Igor L. Shabalin

Ultra-High Temperature Materials II

Refractory Carbides I (Ta, Hf, Nb and Zr Carbides)

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A Comprehensive Guide and Reference Book

Springer

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This book is dedicated to two Russian men:

—Colonel Logan S. Shabalin (1926–2016), one of the Russian soldiers of WWII, my beloved father

and

—Dr. Leonid I. Shabalin (born in 1979), my beloved son

Preface

This book is the second volume in my *Ultra-High Temperature Materials* series, which is devoted to the materials having melting (sublimation or decomposition) points around or above 2500 °C. In the preface to the first volume, I have already detailed all the motives that led me to the creation and establishment of this series of books, so here I consider it is sufficient only to refer to it. The second volume is not only the continuation of the first volume, but taking into account the system of references and addenda adopted in the series, it accumulates all the information presented in both volumes. Thus, any reader/user of this book is unlikely to be satisfied with the process of learning or searching for the information necessary for him without the usage of the previous volume.

Many thanks to the colleagues for their feedback on the chapters of the first volume, but I again ask everybody, who has any remarks, observations, or possibly corrections and personal opinions, concerning this book and its contents, to send them directly to my e-mail. It will be very useful for me to take into account all your responses before preparation and publishing of the succeeding volumes of the series.

This book would not have been possible to create without the permanent encouragement and warm support I have got for the last years from my best friends, my colleagues and co-authors such as Prof. D. Keith Ross, one of the best specialists in the UK in the area of materials characterization, former director of Institute for Materials Research (University of Salford, Manchester), the authors of many books on materials science of refractory compounds, Prof. Rostislav A. Andrievskii (Institute for Problems of Chemical Physics, Russian Academy of Sciences, Chernogolovka, Moscow Region) and Prof. George G. Gnesin (Institute for Problems of Materials Science, National Academy of Sciences of Ukraine, Kyiv). Probably, therefore, it is still difficult for me to imagine that these brilliant scientists, my close friends, recently passed away, are no longer with us.

For the invaluable and kind assistance in the preparation of this volume, I would like gratefully to acknowledge Prof. Alexander I. Savvatimskiy (Joint Institute for High Temperatures, Moscow, Russia), Prof. Irina V. Medvedeva (Institute of Metal Physics, Yekaterinburg, Russia), Prof. Anders E. W. Jarfors (Jönköping University,

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I sincerely apologize if I have missed to mention the assistance of anybody else from the variety of scientists and researchers all around the world I have contacted with concerning the issues connected with the topics of this book.

Lastly, I wish to thank my relatives for their steadfast support during the last 4 years, which were necessary for the complete preparation of the book.

Manchester, UK
November 2017

Professor Igor L. Shabalin

Contents

About the Author

Igor L. Shabalin in his professional career has got over 45 years of experience in Ultra-High Temperature Materials Design, Science and Engineering. He was born in Russia, graduated in Technology of Less-Common Metals and received his M.Sc. and Ph.D. degrees from the Ural Polytechnic University (UPI), Yekaterinburg (former—Sverdlovsk), Russia. He has held academic positions at the UPI and was the founder of the Special Research Laboratory for Aerospace Industry (ONIL-123). As head of the laboratory and member of several scientific and technological councils, he established collaboration between universities and industry by running a variety of R&D projects and was involved in the management of some world leading programmes in rocketry and spacecraft development in the USSR Ministry of Aerospace Industry (MOM). In 2003, he immigrated to the United Kingdom and joined the University of Salford, Manchester, as a researcher in Materials in 2005. As he has developed his personal original approach to a special subclass of engineering materials—hetero-modulus composites and hybrids in ceramics, his research activity focuses mainly on high and ultra-high temperature ceramic composites with graphene-like (carbon and boron nitride) constituents. He has discovered in Russia in the 1980–90s and formulated later in the UK—mesoscopic temperature–pressure-dependent phenomenon in solid-state gas-exchange chemical reactions (surface processes) termed as "ridge effect". From 1971 up to date, he has published about 300 scientific and technical papers and

holds more than 40 patents. In 2014, he was awarded the title of Honorary Professor of the Department of High Temperature Materials (National Technical University of Ukraine), which was founded by Grigorii V. Samsonov, one of the world famous scientists of the twentieth century in the field of physics and chemistry of refractory compounds.

Abstract

The work represents a thorough treatment of ultra-high temperature materials with melting points around or over 2500 °C. The second volume included physical (structural, thermal, electromagnetic, optical, mechanical and nuclear) and chemical (binary, ternary and multicomponent systems, solid-state diffusion, wettability, interaction with chemicals, gases and aqueous solutions) properties of refractory carbide materials: tantalum carbides (monocarbide TaC_{1-x} and semicarbide α/β-$Ta_{2\pm x}C$), hafnium monocarbide HfC_{1-x}, niobium carbides (monocarbide NbC_{1-x} and semicarbide $\alpha/\beta/\gamma$-$Nb_{2\pm x}C$) and zirconium monocarbide ZrC_{1-x}. It will be of interest to researchers, engineers, postgraduate, graduate and undergraduate students alike. The reader/user is provided with the full qualitative and quantitative assessment for the materials, which could be applied in various engineering devices and environmental conditions in the wide range from cryogenic to ultra-high temperatures, on the basis of the latest updates in the field of physics, chemistry, nanotechnology, materials science and engineering.

Chapter 1
Introduction

Referring to the Volume I of this book [1], it is necessary to remind the fact that not only elemental carbon, but also its chemical compounds with various *p*-, *d*- and *f*-elements, termed as carbides, are of great interest for the materials designers and engineers, who are working in the area of ultra-high temperature materials, and so responsible for their technical applications in practice. If you look at the periodic table (*see* Table 1.1), where the melting (or decomposition) points of the binary compounds of chemical elements with carbon (carbides) are marked specially, you will find at least twenty carbide compounds of twelve elements (B, Si, Ti, V, Zr, Nb, Mo, Hf, Ta, W, Th, U), which are solid at the temperatures of around 2500 °C. Some refractory carbides have a bit higher melting points than the accepted decomposition point of carbon, while the most of refractory carbides have much higher melting points than the corresponding metals or non-metals, which form these compounds with carbon.

Namely, the refractory carbides possess the highest melting points among all the solids ever known [1–9]. This amazing property allow some carbides to save their physico-mechanical, thermophysical and physico-chemical stability, including hardness, strength and wear and erosion resistance, at the certain level up to ultra-high temperature ranges. The carbides are also very stable under the conditions of intensive nuclear radiation, when the radiation affects the solids jointly with higher temperatures. The extraordinary and very special behaviour of carbide compounds, formed by *d*- and *f*-elements (transition metals) and also called as metal-like (or interstitial) carbides, is caused by the specific chemical bonds between carbon and metal, which include and combine metallic, ionic and covalent interatomic bonding/interaction simultaneously [10–12]. The heterodesmic character of chemical bonding in transition metal carbides is almost always accompanied by non-stoichiometry and existence of wide homogeneity ranges of these carbide phases (variability of composition), e.g. titanium monocarbide TiC_{1-x} exists in the area of compositions from $TiC_{\sim 0.4}$ to $TiC_{\sim 1.0}$ in the same cubic crystal structure with only certain change of its lattice parameter (*see* section C–Ti in Table I-2.13; hereafter "I-0.00" is used to mark the sections, figures and tables of

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Table 1.1 Melting (or decomposition) points T_m (°C) of the binary compounds of chemical elements with carbon (carbides) in the periodic table [1–10, 13–15]

1/IA	2/IIA	3/IIIB	4/IVB	5/VB	6/VIB	7/VIIB	8/VIIIa	9/VIIIb	10/VIIIc	11/IB	12/IIB	13/IIIA	14/IVA	15/VA	16/VIA	17/VIIA	18/VIIIA
^{1}**H** –																	^{2}**He** –
^{3}**Li** Li_2C_2 (~600)	^{4}**Be** Be_2C (~2100) BeC_2 (?)											^{5}**B** **$B_{4\pm x}C$** **(2450)**	^{6}**C**	^{7}**N** –	^{8}**O** –	^{9}**F** –	^{10}**Ne** –
^{11}**Na** Na_2C_2 (?)	^{12}**Mg** Mg_2C_3 (~700) MgC_2 (~2200)											^{13}**Al** Al_4C_3 (2100)	^{14}**Si** **SiC** **(2545)**	^{15}**P** –	^{16}**S** CS_2 (?) CS (?)	^{17}**Cl** –	^{18}**Ar** –
^{19}**K** K_2C_2 (?)	^{20}**Ca** CaC_2 (~2400)	^{21}**Sc** ScC_{1-x} (2270) Sc_4C_{3+x} (1860) Sc_3C_{4-x} (1800)	^{22}**Ti** **TiC_{1-x}** **(3070)**	^{23}**V** $V_{2\pm x}C$ (2190) V_4C_{3-x} (1320) $V_6C_{5\pm x}$ (~1220) **VC_{1-x}** **(2800)**	^{24}**Cr** $Cr_{23}C_{6\pm x}$ (1610) $Cr_7C_{3\pm x}$ (1770) Cr_3C_{2-x} (1820)	^{25}**Mn** $Mn_{23}C_6$ (1035) $Mn_{15}C_4$ (1020) Mn_3C (~1050) Mn_5C_2 (1170) Mn_7C_3 (1340)	^{26}**Fe** Meta- stable	^{27}**Co** Meta- stable	^{28}**Ni** Meta- stable	^{29}**Cu** –	^{30}**Zn** –	^{31}**Ga** –	^{32}**Ge** –	^{33}**As** –	^{34}**Se** CSe_2 (?)	^{35}**Br** –	^{36}**Kr** –

Table 1.1 (continued)

^{37}**Rb** (?)	^{38}**Sr** SrC_2 (1800)	^{39}**Y** $Y_{2\pm x}C$ (950) $YC_{1\pm x}$ (~2380) Y_4C_5 (1530) Y_3C_{4-x} (~1600) Y_2C_{3-x} (~1500) $YC_{2\pm x}$ (~2380)	^{40}**Zr** **ZrC_{1-x}** **(3450)**	^{41}**Nb** **$Nb_{2\pm x}C$** **(3080)** Nb_4C_{3-x} (1750) $Nb_6C_{5\pm x}$ (~1100) **NbC_{1-x}** **(3600)**	^{42}**Mo** **$Mo_{2\pm x}C$** **(2530)** **Mo_3C_{2-x}** **(2550)** **MoC_{1-x}** **(2600)** MoC (1220)	^{43}**Tc** TcC_{1-x} (?)	^{44}**Ru** –	^{45}**Rh** –	^{46}**Pd** –	^{47}**Ag** –	^{48}**Cd** –	^{49}**In** –	^{50}**Sn** –	^{51}**Sb** –	^{52}**Te** –	^{53}**I** –	^{54}**Xe** –
^{55}**Cs** (?)	^{56}**Ba** BaC_2 (2000)	^{57}**La*** La_2C_{3-x} (1415) $LaC_{2\pm x}$ (2400)	^{72}**Hf** **HfC_{1-x}** **(3950)**	^{73}**Ta** **$Ta_{2\pm x}C$** **(3335)** **Ta_4C_{3-x}** **(2525)** $Ta_6C_{5\pm x}$ (1160) **TaC_{1-x}** **(3990)**	^{74}**W** **$W_{2\pm x}C$** **(2790)** **WC_{1-x}** **(2775)** **$WC_{1\pm x}$** **(2785)**	^{75}**Re** –	^{76}**Os** –	^{77}**Ir** –	^{78}**Pt** –	^{79}**Au** –	^{80}**Hg** –	^{81}**Tl** –	^{82}**Pb** –	^{83}**Bi** –	^{84}**Po** –	^{85}**At** (?)	^{86}**Rn** –
^{87}**Fr** (?)	^{88}**Ra** (?)	^{89}**Ac**** (?)	^{104}**Rf** (?)	^{105}**Db** (?)	^{106}**Sg** (?)	^{107}**Bh** (?)	^{108}**Hs** (?)	^{109}**Mt** (?)	^{110}**Ds** (?)	^{111}**Rg** (?)	^{112}**Cn** (?)	^{113}**Nh** (?)	^{114}**Fl** (?)	^{115}**Mc** (?)	^{116}**Lv** (?)	^{117}**Ts** (?)	^{118}**Og** (?)

Table 1.1 (continued)

*	^{58}**Ce** Ce_2C_3 (~1600) $CeC_{2\pm x}$ (2300)	^{59}**Pr** Pr_2C_{3-x} (1550) $PrC_{2\pm x}$ (2320)	^{60}**Nd** Nd_2C_{3-x} (1620) NdC_2 (2260)	^{61}**Pm** (?)	^{62}**Sm** $Sm_{3-x}C$ (?) Sm_2C_{3-x} (1325 ?) SmC_2 (~2300)	^{63}**Eu** $Eu_{3-x}C$ (?) Eu_2C_3 (?) EuC_2 (1700 ?)	^{64}**Gd** $Gd_{3-x}C$ (?) Gd_2C (?) Gd_2C_{3-x} (?) GdC_2 (2370)	^{65}**Tb** $Tb_{3-x}C$ (?) Tb_2C_{3-x} (?) TbC_2 (~2150)	^{66}**Dy** $Dy_{3-x}C$ (?) Dy_2C (?) Dy_2C_{3-x} (?) DyC_2 (2250 ?)	^{67}**Ho** $Ho_{3-x}C$ (?) Ho_2C (?) Ho_2C_{3-x} (?) HoC_2 (2270 ?)	^{68}**Er** $Er_{3-x}C$ (?) $Er_{15}C_{19}$ (?) ErC_2 (2250 ?)	^{69}**Tm** $Tm_{3-x}C$ (?) $Tm_{15}C_{19}$ (?) TmC_2 (2180 ?) TmC_3 (?)	^{70}**Yb** $Yb_{3-x}C$ (?) YbC_{1-x} (?) Yb_4C_{5+x} (?) $Yb_{15}C_{19}$ (?) $Yb_3C_{4\pm x}$ (?) YbC_2 (?)	^{71}**Lu** $Lu_{3-x}C$ (?) $Lu_{15}C_{19}$ (?) LuC_2 (?)
**	^{90}**Th** **$ThC_{1\pm x}$** **(2630)** **ThC_{2-x}** **(2660)**	^{91}**Pa** PaC (?) PaC_2 (?)	^{92}**U** **$UC_{1\pm x}$** **(2530)** U_2C_3 (1820) **UC_{2-x}** **(2510)**	^{93}**Np** NpC_{1-x} (?) Np_2C_3 (?) NpC_2 (?)	^{94}**Pu** Pu_3C_2 (575) PuC_{1-x} (1655) Pu_2C_{3-x} (2040) PuC_2 (2230)	^{95}**Am** Am_2C_3 (?) AmC_2 (?)	^{96}**Cm** (?)	^{97}**Bk** (?)	^{98}**Cf** (?)	^{99}**Es** (?)	^{100}**Fm** (?)	^{101}**Md** (?)	^{102}**No** (?)	^{103}**Lr** (?)

Volume I of this book). Non-metallic or covalent carbides have also great resistance to the outer impact of high-temperature environment, but differ in kind of their chemical bonding from the metal-like carbides significantly, although non-stoichiometry might be inherent to their structures as well [13–15]. These both types of carbon compounds: interstitial (metal-like) and covalent carbides, according to the classification of inorganic compounds of carbon given earlier (*see* section I-2.6), can be used as materials or constituents of composite materials for high and ultra-high temperature technical applications.

The classification of refractory carbides into metallic and non-metallic is very important, since it is directly related to the physics of the transport properties of carbides. Naturally, the electrical and thermal conductivities of the former have a metallic character, while non-metallic refractory carbides are characterized mainly by semiconductor properties. A correct understanding of both the nature of refractory carbides and the manufacturing processes of materials on their basis allows material designers to obtain for carbide compositions the various necessary for practice temperature dependences of conductivities [16].

The refractory carbides vary sufficiently in their possible structures and compositions and physical and chemical properties. Some of them are exceedingly inert to chemical attacks, though their chemical stability has commonly a selective character. At higher temperatures all the carbides are susceptible to gas corrosion in oxidizing atmospheres, although some carbides have better resistance to oxidation than do the refractory metals, and most of them are more resistant to oxidation than graphite or other carbonaceous materials. A thorough consideration should be given to the atmosphere, in which refractory carbides are employed, as some of them can be attacked by reducing gas atmosphere at higher temperatures as well [17]. In order to realize successfully the high and ultra-high temperature properties of carbides, special attention must be paid first of all to the issues of their chemical compatibility in contact with various solid, liquid and gaseous substances in wide range of temperatures and pressures of the medium.

In addition to the extremely high melting points, carbides possess a wide range of other valuable properties, among which, in the first place, it is necessary to note their super-high hardness [18]. For massive machinery production all around the world, this property of carbides is much more important than anything else. So, for the reader/users, who are interested in ambient temperature applications of hard carbides, this book on ultra-high temperature materials will be interesting too. The fundamental significance of refractory carbides lies in the fact that in their lattices the breakdown of sp^3-configurations has a considerable extent deferred to much higher temperatures than those at which it occurs in diamond, “champion of hardness”, in whose lattice there are no such stabilizers of electron configurations, as for example titanium or silicon in corresponding carbides. Thus, refractory carbides represent a unique compromise as their hardness, although less than that of diamond, is retained at the temperatures, which are unreachable in general for such material as diamond [19].

In this volume for each individual carbide compound or for a metal carbide family (various carbides of the same metal) being considered, as it was before for

the chemical elements (carbon and metals) in Volume I, a special section devoted to its structural features (partial variants of metal-carbon phase diagram, system and type of crystal structures, composition—lattice parameter relationships, possible phase transformations and slip systems, effects of contamination on structures, synthesized or theoretically predicted nanostructures, density of structures) forestalls the thorough and comprehensive description of data available in the literature on thermal, electro-magnetic and optical, physico-mechanical and nuclear physical properties of this carbide (or carbide family) with various compositions (carbon-metal ratios) within its homogeneity range and over very wide temperature range from 0 up to 2700–5000 K. The chemical properties of carbides (or carbides families) are considered in special sections by the analysis of interaction of carbides with metals, non-metals, various refractory compounds (carbides, nitrides, oxides, borides, silicides, intermetallides), gases and some common chemicals (acids, alkalies and salts in aqueous solutions and melts), diffusion processes in carbides and their wettabilities by various melts in the wide range of temperatures. As it was before in Volume I, the most reliable data (in author's opinion) on main physical properties of the ultra-high temperature carbide materials included in this volume are summarized in Addendum. Indeed, all the data in Addendum are based on the information given before in the corresponding chapters; however, every reader/user of the book has to be reminded that these data relate only to the quasi-stoichiometric compositions (or carbon-richest compositions for non-stoichiometric vanadium monocarbide) of the corresponding carbide materials.

The full amount of data on the structures and properties of refractory carbides, which is available in literature at the moment of the publication of this reference book was turn up to be too enormous, so it has been decided to divide all the information on ultra-high temperature carbide materials into three volumes. Honestly, it was one of the kind wishes of my readers/users, who took advantage of the feedback, to make the real content of the book more "deep" than "wide". Thus, in this volume anyone can find all the necessary information on the transition metal carbides of 4–5 groups of the periodic table having the highest melting points (in the order of decreasing melting points): tantalum carbides (monocarbide TaC_{1-x} and semicarbide α/β-$Ta_{2\pm x}C$), hafnium monocarbide HfC_{1-x}, niobium carbides (monocarbide NbC_{1-x} and semicarbide $\alpha/\beta/\gamma$-$Nb_{2\pm x}C$) and zirconium monocarbide ZrC_{1-x}. As far as possible, due to the reduction in the number of systems under the consideration, the author has tried to include in this volume all the latest research data on nanostructures of these refractory carbides, as well as the results of computer simulation for the properties of these carbides based on the quantum chemical and other theoretical approaches and calculations. In Table 1.1, all the refractory carbide systems relegated according to the classification adopted in this book to the ultra-high temperature materials are shaded gray; those carbide systems, which are considered in the second volume, are marked out in bold additionally. The next volumes of the reference book series will include data on metallic carbide systems, such as titanium, vanadium, tungsten, molybdenum, thorium and uranium carbides, as well as non-metallic refractory carbide systems, such as silicon and boron carbides.

There are no special sections devoted to the manufacturing methods of refractory carbides in the book chapters; however, anybody can find a lot of useful information on this issue in the sections, where the chemical properties of carbides and general principles of carbide containing materials design are described. The special author's remarks for readers/users of the book are following:

when it is not indicated specially, the value of percentage reported in the text of the book is given in mass percents;
some experimental and/or theoretically calculated data presented in the text, which are a bit doubtful in author's opinion, as they have not been confirmed in the literature sufficiently, are denoted by the special question marks.

Finally, it is necessary to recommend all the readers/users to become acquainted with the introduction of the previous volume [1], as those novel approaches to the description of materials that were substantiated there are also used by the author in this volume of book series.

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Chapter 2
Tantalum Carbides

2.1 Structures

Tantalum forms with carbon several chemical compounds (*see* also section C–Ta in Table I-2.13): tantalum monocarbide TaC_{1-x} with the extremely broad homogeneity range, low-temperature (ordered) α-$Ta_{2+x}C$ and high-temperature (disordered) β-$Ta_{2\pm x}C$ modifications of tantalum semicarbide (the latter cannot be retained by quenching), close-packed (nanolamellar) ε-$Ta_3C_{2\pm x}$ and ζ-Ta_4C_{3-x} phases, also proposed $Ta_5C_{4\pm x}$, $Ta_6C_{5\pm x}$, $Ta_7C_{6\pm x}$ and $Ta_8C_{7\pm x}$ ordered phases and "surface" phase Ta_5C_2 [1–3, 17, 21, 76, 440, 679, 707, 730, 731]. Additionally, the absorption of a small amount of carbon was said to result in the ordered subcarbide $Ta_{64}C$ [26, 528], which is more likely an example of a phase subject to controversies [27]. A high-temperature partial variant of tantalum – carbon phase diagram is given in Fig. 2.1, and the structural features of the tantalum carbides are presented in Table 2.1. The C/Ta atomic radii ratio calculated on the basis of Pauling's atomic size of Ta (0.1457 nm, CN = 12) is 0.529 [115], 0.57 [90]; the ratio of Ta radii (in nm) in Me/MeC is 0.142/0.157 (10.5% expansion of Ta atoms in carbide) [155]. The structural transformation of TaC_{1-x} from NaCl type (*Fm(–3)m*) to CsCl type (*Pm(–3)m*) structure under ultra-high pressures (~100–700 GPa) has been predicted theoretically [410–412, 431, 706, 716].

In contrast to the other transition metal refractory carbides, within the homogeneity range TaC_{1-x} demonstrates practically a linear relationship (some kinds of exception are the data obtained by Santoro [122], Senczyk [399] and Gusev [754]) between lattice parameter and composition (*see* Fig. 2.2); in modified forms the following equations, earlier proposed by Bowman (for $0.01 \le x \le 0.29$) [78]:

$$a = 0.4457 - 0.0156x, \tag{2.1}$$

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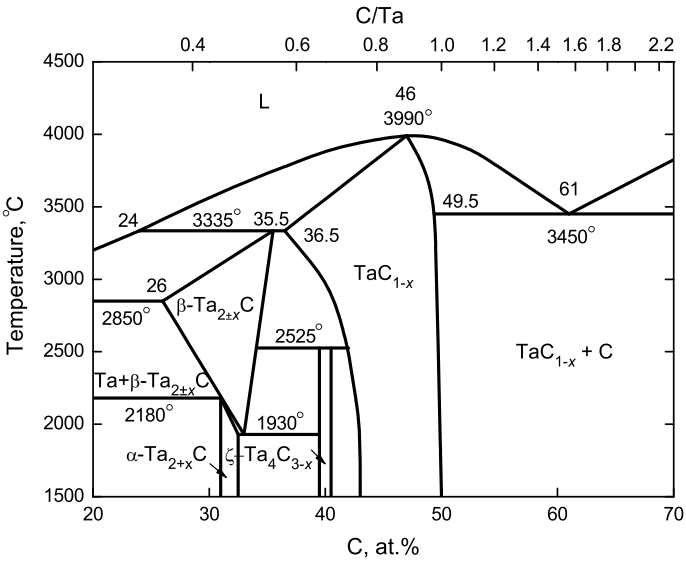

Fig. 2.1 High-temperature partial variant of tantalum – carbon equilibrium phase diagram [2, 4, 5, 80, 188, 463–468, 653, 757, 811]

later by Storms (for $0 < x \leq 0.28$) [17]:

$$a = 0.4458 - 0.0167x, \tag{2.2}$$

and then by Alexandre et al. [398] for materials (with $0 < x \leq 0.15$) reactively processed by hot isostatic pressing (HIP):

$$a = 0.4454 - 0.0102x, \tag{2.3}$$

can be recommended to describe it, where a is lattice parameter, nm and x is the value of index in TaC_{1-x} formula. However, in the contradiction to the linear relationships for the same correlation the complex (polynomial) functions were suggested by Senczyk ($0 \leq x \leq 0.30$) [399]:

$$a = 0.4457 - 0.0127x - 0.0130x^2 \tag{2.4}$$

and Gusev ($x \leq 0.329$) [754]:

$$a = 0.4456 - 0.0159x + 0.0077x^2. \tag{2.5}$$

Within the homogeneity range of semicarbide phase the lattice parameter a also declines with carbon content decreasing [16]. For parameters of formation and migration of lattice point defects (vacancies and interstitial atoms) in tantalum monocarbide *see* Sect. 2.5 (Table 2.19). At low temperatures for non-stoichiometric

Table 2.1 Structural properties (crystal structure, density) of tantalum carbide phases

Formula	Crystal structure						Z[b]	Density[c], g cm^{-3}	References
	System	Type	Space group	Lattice parameters[a], nm					
				a	*c*	*c*/*a*			
TaC_{1-x}	Cubic	NaCl	*Fm(–3)m*	0.4410[d]	–	–	4	–	[17, 37, 78]
				0.4424[e]	–	–	4	–	[201]
				0.4427	–	–	4	14.60	[9]
				0.4438	–	–	4	14.57	[12]
				0.4439[f]	–	–	4	–	[758]
				0.4445	–	–	4	–	[6]
				0.4448	–	–	4	–	[10]
				0.4450	–	–	4	–	[43]
				0.4451	–	–	4	–	[14]
				0.4452	–	–	4	14.53	[49]
				0.4453[g]	–	–	4	14.52	[8, 34, 46]
				0.4454	–	–	4	14.48	[1, 13, 35, 123, 159]
				0.4455	–	–	4	14.48	[11, 16, 17, 41, 50, 60]
				0.4456[h]	–	–	4	14.48	[15, 17, 48]
				0.4460	–	–	4	14.45	[7]
β-$Ta_{2\pm x}C$[i]	Hexagonal	NiAs	*P6$_3$/mmc*	0.3102	0.4940	1.593	1	–	[41]
				0.3105	0.4935	1.589	1	15.03	[21]
				0.3105	0.4945	1.593	1	–	[24, 25]
				0.3108	0.4950	1.593	1	15.00	[23]
				0.3119	0.4934	1.582	1	–	[32]
α-$Ta_{2+x}C$[i,k]	Trigonal	CdI_2[l]	*P(–3)m1*	0.3094	0.4918	1.590	1	–	[449]
				0.3097	0.4930	1.592	1	15.16	[6]
				0.3100[m]	0.4931[m]	1.590[m]	1	–	[21, 22]
				0.3102[n]	0.4940[n]	1.593[n]	1	–	[21, 22, 77]
				0.3103	0.4936	1.591	1	–	[36]
				0.3103	0.4937	1.591	1	15.08	[19]
				0.3103[n]	0.4938[n]	1.591[n]	1	15.08	[20, 21]
				0.3104	0.4927	1.587	1	–	[34]
				0.3104[o]	0.4939[o]	1.591[o]	1	–	[33, 97]
				0.3104	0.4940	1.591	1	15.04	[1, 35]
				0.3104	0.4941	1.591	1	–	[43, 543]
				0.3106	0.4945	1.592	1	–	[16, 17, 41]
				0.3106	0.4964	1.593	1	–	[18]
				0.3111	0.4948	1.590	1	–	[449]
ε-$Ta_3C^{p}_{2\pm x}$	Trigonal	TiS	*R(–3)m*	0.3110	2.240	7.203	3	–	[3]
ζ-$Ta_4C^{p}_{3-x}$	Trigonal	$Sc_2Te^{p}_3$	*R(–3)m*	0.3116	3.000	9.628	3	15.01	[3, 28]
				0.3120	3.001	9.620	3	–	[34]
				0.3122[q,r]	3.006[q,r]	9.629[q,r]	3	14.84	[29]
				0.3123[r]	3.005[r]	9.623[r]	3	–	[32]
$Ta_6C^{s,t}_{5\pm x}$	Mono-clinic or trigonal	–	*C2/m*, *C2/c*, *P3$_1$*, *P3$_1$12*	–	–	–	–	–	[2, 30, 31, 730]

(continued)

Table 2.1 (continued)

[a]When it is not indicated specially, value reported is for near-stoichiometric composition
[b]Number of formula units per lattice cell
[c]Calculated from XRD or neutron diffraction patterns
[d]Carbon content – 41.2 at.% (in equilibrium with semicarbide phase)
[e]Calculated on the basis of density-functional theory (DFT)
[f]Nanocrystalline powders, specific surface area – 25 $m^2\ g^{-1}$, mean particle size – 15-25 nm, carbon content – 49.0 at.%, oxygen content – 0.3-0.4 mas.%
[g]Minimum interatomic distances: Ta–Ta and C–C – 0.314875 nm, Ta–C – 0.222650 nm [46]
[h]Carbon content – 49.7 at.% (in equilibrium with graphite)
[i]In contradiction to these data, β-$Ta_{2\pm x}C$ was identified as orthorhombic (ζ-Fe_2N, *Pbcn*) with a = 0.4940 nm, b = 0.6208 nm, c = 0.5370 nm, Z = 4 [1], or orthorhombic (?) with a = 0.4933 nm, b = 0.9866 nm, c = 0.8534 nm [37]
[j]In contradiction to these data, α-$Ta_{2+x}C$ was identified as orthorhombic (?) with a = 0.2873 nm, b = 1.0250 nm, c = 0.4572 nm [60]
[k]Also structured (vacancy ordered) in *Pnma*, *Pbcn*, *P3m1*, *I4₁/amd*, *R(–3)m*, *Fd(–3)m* and *P4/mmm* [730]
[l]Cadmium iodide antitype structure
[m]Carbon content – 31.5 at.%
[n]Carbon content – 33 at.% C
[o]Carbon content – 32-33 at.% C
[p]In contradiction to these data, $Ta_4C_{3.04}$ ($TaC_{0.76}$) was identified as cubic (*Pm(–3)m*) with a = 0.4424 nm, Z = 1 and calculated density – 14.58 g cm^{-3} [47]
[q]Also $GeSb_2Te_4$ structure type proposed [29]
[r]Carbon content – 40 at.% C
[s]Incommensurate superstructure close to M_6C_5 type [30, 31]
[t]Also vacancy ordered $Ta_5C_{4\pm x}$ (*P(−1)*, *C2/m*, *I4/m*), $Ta_7C_{6\pm x}$ (*R(−3)*) and $Ta_8C_{7\pm x}$ phases (*P4₃32*, *Fm(–3)m*) are supposed [730]

monocarbide TaC_{1-x} the slip system (110) <110> was determined, while for quasi-stoichiometric monocarbide the slip system (111) <110> is preferential at all temperatures [154–157, 159, 182, 381, 439, 517, 525, 533]; at temperatures above ~1200 °C slip can occur on other planes, e.g. in the systems (001) <010>, (011) <010> and (011) <110> [157, 381, 533]. At ambient temperatures Kiani et al. [564, 720] employing *in situ* electron microscopy method observed for $TaC_{\sim 0.85}$ single crystalline nanopillars the $(1\bar{1}0)$ <110> and $(1\bar{1}1)$ <110> slip systems; the DFT-calculated by them energy values, which characterize several different possible slip systems in quasi-stoichiometric TaC_{1-x}, are given in Table 2.2. The minimal Burgers vector of near-stoichiometric TaC_{1-x} (½ <110>) b = 0.313 nm [41]. Theoretically for α-$Ta_{2+x}C$ phase with its trigonal symmetry easier slip would occur in (0001) $<11\bar{2}0>$ (on the basal plane); however $(10\bar{1}1)$ $<11\bar{2}3>$ (on a pyramidal plane) dislocation slip was confirmed experimentally [77, 189, 517]. The variations of the lattice parameter of TaC_{1-x} with temperature for the phases having different deviations from the stoichiometry are shown in Fig. 2.3.

In general, the contaminations, such as dissolved in TaC_{1-x} oxygen or/and nitrogen, which are always present in the materials, decrease the magnitude of lattice parameter by forming oxycarbide $TaC_{1-x}O_y$, carbonitride $TaC_{1-x}N_z$ and oxycarbonitride $TaC_{1-x}N_zO_y$ phases with various deviations from the stoichiometry

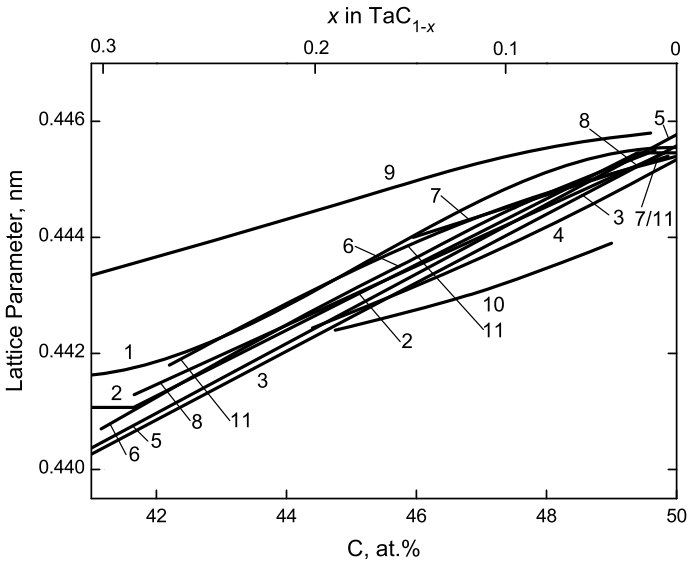

Fig. 2.2 Lattice parameter of TaC_{1-x} as a function of phase composition: 1 – [122], 2 – [17], 3 – for disordered structures [30, 754–756], 4 – [34], 5 – polynomial approximation [35, 123], 6 – [399], 7 – [398], 8 – synthesized by different methods [332, 753], 9 – [207], 10 – nanocrystalline powders (specific surface area – ~25 $m^2\ g^{-1}$, mean particle size – 10-30 nm, content O – 0.3-0.4%) [758, 759], 11 – synthesized from the pure elements [604, 651, 766] (2-3, 5, 8 – on the basis of several sources)

Table 2.2 The DFT-calculated values of unstable stacking fault energies and total energies required to shear for the crystals of quasi-stoichiometric tantalum monocarbide TaC_{1-x} along different slip systems [564]

Slip system	Maximum E_{GSF} (l)[a], eV nm^{-2}	Shear total energy, eV nm^{-3}
(0 0 1) <110>	13.9	92
(1 $\bar{1}$ 0) <110>	10.3	492
(1 $\bar{1}$ 1) <110>	3.5	476
(0 0 1) <100>	16.8	–
(1 $\bar{1}$ 0) <100>	30.9	–
(1 $\bar{1}$ 0) <111>	33.6	–

[a]The generalized stacking fault energy E_{GSF} is a function of normalized displacement (l/b), where b is the Burgers vector for the slip direction; so E_{GSF} (l) is defined as the energy per unit area required to shift one half of the supercell with respect to the other half by a distance l

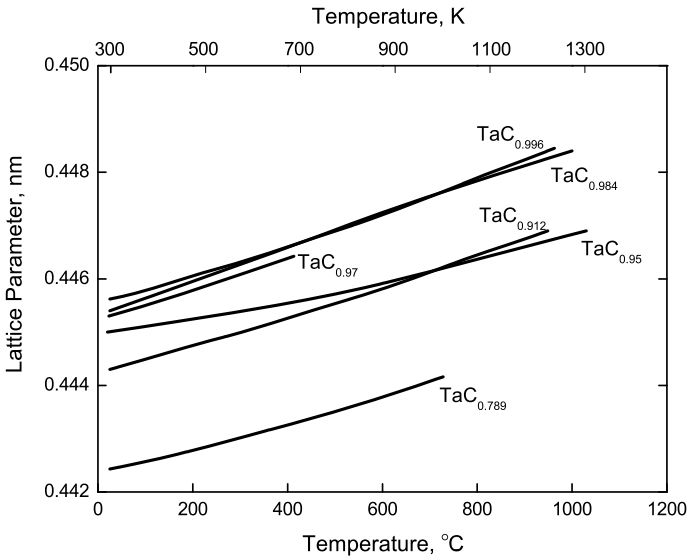

Fig. 2.3 Lattice parameter of TaC_{1-x} as a function of temperature for the phases having different deviations from the stoichiometry ($x = 0.004$ [48], $x = 0.016$ [79], $x = 0.03$ [46], $x = 0.05$ [43, 86], $x = 0.088$ and $x = 0.211$ [48])

(*see* also sections C–N–Ta and C–O–Ta in Table I-2.14); the effect of namely oxygen in this case is inconsiderable as its solubility in TaC_{1-x} is much lower in comparison with nitrogen [36, 63, 226, 530, 570].

Recently, due to the massive progress in nanotechnology, various nanostructures based on TaC_{1-x}, including nanorods (14–40 nm diameter, 800 nm length), nanowires, nanofibers (~40–90 nm in diameter, up to several micrometers in length), whiskers, thin films (300 nm thickness) and various (spherical, cubic-faceted) nanoparticles (40–250 nm), have been prepared [73, 74, 117, 450, 693, 702, 717–719, 758–760, 767–770, 816], as well as 2D-molecular $Ta_{n+1}C_n$ MXenes (Ta_2C, Ta_3C_2, Ta_4C_3, or $Ta_4C_3T_x$, where T are functional groups OH, O, F) – were synthesized and investigated in details theoretically [75, 563, 698, 699, 710, 715, 727, 749–751].

First Bowman [78] – for the X-ray density of synthesized from the elements TaC_{1-x} ($0.01 \le x \le 0.29$), and later Samsonov and Petrykina [705] – for the bulk density of hot-pressed TaC_{1-x} materials ($0 < x \le 0.20$), proposed the equation:

$$d = 14.47 + 0.64x, \tag{2.6}$$

where d is the density, g cm^{-3} and x is the value of index in TaC_{1-x} formula. According to Storms [17], the X-ray density for monocarbide TaC_{1-x} phase with $x = 0.01$ is 14.48 g cm^{-3} and it increases as carbon is removed. The recommended values for the bulk (or pycnometric) density of pure poreless materials at room temperature are 14.8–15.0 and 14.4–14.6 g cm^{-3} – for the near-stoichiometric single-phase compositions of tantalum semicarbide $Ta_{2\pm x}C$ and monocarbide TaC_{1-x}, respectively [16, 37, 43, 59–61, 388, 442, 544, 701, 721].

2.2 Thermal Properties

Tantalum monocarbide TaC_{1-x} has the highest melting point of all the solid substances available; the incongruent melting point of β-$Ta_{2\pm x}C$ is about 500° lower. Within the homogeneity range of TaC_{1-x} the melting point of the phase varies (*see* Fig. 2.1); the maximum temperatures pertain to the nonstoichiometric compositions around ~$TaC_{0.88–0.89}$ [17, 70, 71], and there are no sufficient proofs for complex carbides to be outperforming the maximum melting point of pure tantalum monocarbide phase [71]. The general thermodynamic properties of near-stoichiometric tantalum carbide phases are summarized in Tables 2.3 and 2.4. For the molar heat capacity of near-stoichiometric TaC_{1-x} $c_p = f(T, \mathrm{K})$, J mol^{-1} K^{-1}, the following relationships were recommended in the literature:

in the range of temperatures from 300 to 4000 K [726]

$$c_p = 43.30 + \left(8.16 \times 10^{-3}\right)T - \left(7.95 \times 10^{5}\right)T^{-2}, \tag{2.7}$$

in the range of temperatures from 300 to 2000 K [44, 52]

$$c_p = 30.46 + \left(6.904 \times 10^{-3}\right)T, \tag{2.8}$$

in the range of temperatures from 1600 to 2300 K [52, 58]

$$c_p = 46.36 + \left(6.643 \times 10^{-3}\right)T, \tag{2.9}$$

in the range of temperatures from 1300 to 2850 K [52, 55]

$$c_p = 47.72 + \left(5.532 \times 10^{-3}\right)T, \tag{2.10}$$

in the range of temperatures from 1000 to 3000 K [52, 56]

$$c_p = 44.60 + \left(6.430 \times 10^{-3}\right)T - \left(7.606 \times 10^{5}\right)T^{-2}, \tag{2.11}$$

in the range of temperatures from 370 to 3525 K [52, 57]

$$c_p = 46.02 + \left(3.714 \times 10^{-3}\right)T - \left(2.826 \times 10^{5}\right)T^{-2} + \left(6.620 \times 10^{13}\right)T^{-2} \times \exp\left(-39000/T\right). \tag{2.12}$$

The variations of molar heat capacity with temperature for near-stoichiometric tantalum carbides $Ta_{2\pm x}C$ and TaC_{1-x} are demonstrated in Fig. 2.4 on the basis of several sources. For the molar heat capacity of monocarbide TaC_{1-x} phases with various deviations from the stoichiometry the following relationships were recommended:

Table 2.3 General thermodynamic properties of near-stoichiometric tantalum monocarbide TaC_{1-x}

Characteristics	Symbol	Unit	Value	References
Standard heat of formation (at 298.15 K)[a]	$-\Delta H^\circ_{298}$	kJ mol^{-1}	142.4	[17]
			142.6 ± 6.0[b]	[2, 818–820]
			142.7[c]	[60, 578]
			143.4	[361, 362]
			144.1	[38]
			144.8	[1, 16, 39]
			144.9 ± 3.8	[37]
			146.4	[80, 205]
			150.0	[61]
			159.5	[35, 555]
			161.2	[43, 545]
Standard molar entropy[d] (at 298.15 K and 100 kPa)	S°_{298}	J mol^{-1} K^{-1}	41.9	[61]
			42.26	[362]
			42.29 ± 0.84	[16, 17, 37, 43]
			42.34	[1]
			42.38	[38, 52, 53]
			43.40 ± 0.33	[54]
Molar enthalpy difference	$H_{298}-H_0$	kJ mol^{-1}	6.460	[52, 53]
Standard molar heat capacity[e] (at 298.15 K and 100 kPa)	$c^\circ_{p,298}$	J mol^{-1} K^{-1}	32.50	[44]
			32.52	[361]
			33.75[f]	[360]
			33.89[g]	[360]
			34.81[h]	[360]
			35.42[i]	[360]
			35.71[j]	[360]
			35.82[k]	[360]
			36.40	[60]
			36.50	[61]
			36.69	[17, 41]
			36.78	[1, 16, 54, 362]
			36.80	[38, 43, 52, 53]
Specific heat capacity (at 298.15 K)	c	J kg^{-1} K^{-1}	188.0	[43]
			190.6	[16]
Molar enthalpy (heat) of melting (at the melting point)	ΔH_m	kJ mol^{-1}	105	[16, 579]
Specific enthalpy (heat) of melting (at the melting point)		kJ kg^{-1}	544	[16]

(continued)

Table 2.3 (continued)

Characteristics	Symbol	Unit	Value	References
Melting point	T_m	K (°C)	4050 (3780)	[545–549]
			4070 (3800)	[44]
			4150 ± 150 (3880 ± 150)	[43, 59, 61, 66, 362, 388]
			4220 (3950)	[60]
			4235 (3965)	[576]
			4240 (3970)	[80]
			4250 (3980)[l]	[42, 554]
			4255 ± 50 (3985 ± 50)	[1, 22, 37, 41, 115, 550–553]
			4260 ± 40 (3990 ± 40)[m]	[4, 577]
			4270 ± 200 (4000 ± 200)[n]	[16, 17, 45, 70, 71, 700]
			4420 (4150)	[442]
Boiling point	T_b	K (°C)	5000–5100 (4730–4830)	[388]
			5050 (4780)	[59]
			5770 (5500)	[43, 44, 362]

[a]Enthalpy (heat) of complete dissociation (atomization) from solid state at 298.15 K ($-\Delta_{at}H°_{298}$, kJ mol^{-1}): 1642 ± 9 [1, 820], 1651 [572, 824], 1658 [573], 1987 [574]
[b]Extrapolated to the stoichiometric composition of monocarbide
[c]In kJ per g-atom Ta
[d]Molar entropy $S°_T$ (at 1200 K), J mol^{-1} K^{-1}, for TaC_{1-x} phases with different deviations from the stoichiometry: 109.42 (x = 0), 107.40 (x = 0.05), 105.36 (x = 0.10), 103.29 (x = 0.15), 101.20 (x = 0.20), 99.08 (x = 0.25) and 96.94 (x = 0.30) [52]
[e]Molar heat capacity $c_{p,T}$ (at 1200 K), J mol^{-1} K^{-1}, for TaC_{1-x} phases with different deviations from the stoichiometry: 52.63 (x = 0), 51.03 (x = 0.05), 49.41 (x = 0.10), 47.77 (x = 0.15), 46.11 (x = 0.20), 44.43 (x = 0.25) and 42.73 (x = 0.30) [52]
[f]Disordered structure with carbon content – 44.8 at.%
[g]Ordered structure with carbon content – 44.8 at.%
[h]Disordered structure with carbon content – 45.7 at.%
[i]Ordered structure with carbon content – 45.7 at.%
[j]Disordered structure with carbon content – 46.8 at.%
[k]Ordered structure with carbon content – 46.8 at.%
[l]Carbon content – 46.2-47.4 at.%
[m]Carbon content – 47 at.% [577]
[n]Carbon content – ~44.4-47.4 at.% [17, 70, 71]

Table 2.4 General thermodynamic properties near-stoichiometric tantalum semicarbide $Ta_{2\pm x}C$

Characteristics	Symbol	Unit	Value	References
Standard heat of formation (at 298.15 K)[a]	$-\Delta H°_{298}$	kJ mol^{-1}	193.6	[39]
			195.0 ± 8.4[b]	[64]
			197.5	[16]
			197.6 ± 14.2	[37]
			200.4	[83]
			201.0	[17]
			202.6	[80]
			204.2	[84]
			208.4	[1, 60, 578]
			209.0[c]	[64]
			209.1	[62]
			210.0 ± 8.4[d]	[64]
			213.4	[38, 545]
Standard molar entropy (at 298.15 K and 100 kPa)	$S°_{298}$	J mol^{-1} K^{-1}	81.59	[1, 16]
			83.30	[43]
			83.32	[37]
			83.68	[38]
Standard molar heat capacity (at 298.15 K and 100 kPa)	$c°_{p,298}$	J mol^{-1} K^{-1}	60.95	[1, 38]
			60.96	[16]
Specific heat capacity (at 298.15 K)	c	J kg^{-1} K^{-1}	162.0	[43]
			163.1	[16]
Molar enthalpy (heat) of vaporization (at 298.15 K)[e]	ΔH_v	kJ mol^{-1}	2483 ± 4	[575]
Melting point	T_m	K (°C)	3600 ± 30 (3330 ± 30)	[1, 17, 37, 41, 59, 60, 576]
			3610 ± 30 (3335 ± 30)	[4]
			3670 (3400)	[44, 362]
			3770 (3500)	[16, 38]
Boiling point	T_b	K (°C)	5740 (5465)	[576]

[a]Enthalpy of complete dissociation (atomization) from solid state at 298.15 K ($-\Delta_{at}H°_{298}$): 2475 ± 24 kJ mol^{-1} [1]
[b]For $TaC_{0.455\pm0.002}$ composition
[c]Interpolated to $TaC_{0.5}$ composition
[d]For $TaC_{0.507\pm0.002}$ composition
[e]Due to an insignificant change in the phase composition of $Ta_{2.0}C$ during congruent vaporization, the calculation was carried out in accordance to the third law of thermodynamics with the assumption of phase composition constancy

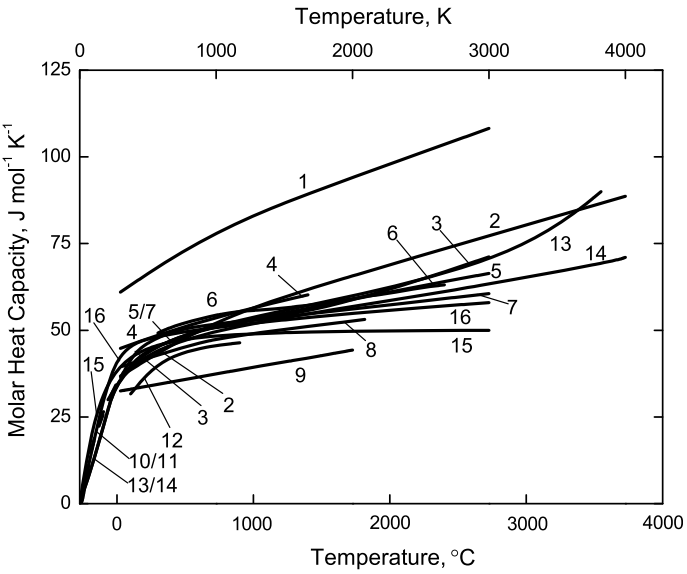

Fig. 2.4 Variation of molar heat capacity (1–14 – at constant pressure, c_p, and 15–16 – at constant volume, c_v) with temperature for near-stoichiometric tantalum carbides $Ta_{2\pm x}C$ (1 – [16]) and TaC_{1-x} (2 – [16, 81], 3 – [17, 209], 4 – [729], 5 – [80], 6 – [51], 7 – [82], 8 – $TaC_{0.99}$ [391, 392], 9 – [44], 10 – [43], 11 – [360], 12 – [771]; theoretically calculated: 13 – Slater's approximation and 14 – Dugdale-MacDonald's approximation on the basis of Debye-Grüneisen model [209], 15 – using projector augmented-wave (PAW) method within the generalized gradient approximation (GGA) of the Perdew-Burke-Ernzerhof scheme for ionic solids (PBEsol) [617], 16 – using density functional theory (DFT) within the GGA with the Perdew-Wang parameterization (PW91) [413]; 1-3, 5-7, 9-10 – on the basis of several sources)

for $TaC_{0.70}$ (in the range of temperatures from 1200 to 2300 K) [40, 52]

$$c_p = 37.46 + \left(5.698 \times 10^{-3}\right)T, \tag{2.13}$$

for $TaC_{0.78}$ (in the range of temperatures from 1200 to 2300 K) [40, 52]

$$c_p = 38.07 + \left(6.770 \times 10^{-3}\right)T, \tag{2.14}$$

for $TaC_{0.85}$ (in the range of temperatures from 1200 to 2300 K) [40, 52]

$$c_p = 38.41 + \left(8.158 \times 10^{-3}\right)T, \tag{2.15}$$

for $TaC_{0.99}$ (in the range of temperatures from 300 to 3000 K) [17]

$$c_p = 57.74 - \left(9.20 \times 10^{-3}\right)T + \left(46.39 \times 10^{-7}\right)T^2 - \left(16.64 \times 10^5\right)T^{-2}, \tag{2.16}$$

for $TaC_{0.99}$ (in the range of temperatures from 1200 to 2300 K) [40, 52]

$$c_p = 39.26 + \left(10.134 \times 10^{-3}\right)T \tag{2.17}$$

and for TaC_{1-x} in general (in the range of temperatures from 1200 to 2300 K) [52]

$$c_p = 41.40 + \left(9.356 \times 10^{-3}\right)T - 31.74x - 4.15x^2, \tag{2.18}$$

or

$$c_p = 30.54 - \left(6.280 \times 10^{-3}\right)T + \left[3.234 + \left(6.176 \times 10^{-3}\right)T\right] \times e^{1-x}, \tag{2.19}$$

where x is the value of index in TaC_{1-x} formula. It should be emphasized specially that all the thermal properties of tantalum carbides are very sensitive to the deviation from the stoichiometry in the phases: for the standard heat of formation $\Delta H°_{298}$, kJ mol^{-1}, molar enthalpy difference $H°_T - H°_{298}$, J mol^{-1}, and molar entropy $S°_T$, J mol^{-1} K^{-1}, the following concentration and temperature-concentration dependencies within the homogeneity range of TaC_{1-x} were obtained by the calculations based on experimental data [2, 35, 39, 40, 52, 62, 818–820]:

$$\Delta H^\circ_{298} = 93.43x - 141.04, \tag{2.20}$$

$$\Delta H^\circ_{298} = 88.55x - 148.38, \tag{2.21}$$

$$\Delta H^\circ_{298} = 95.40 - 434.2(1-x) + 196.1(1-x)^2, \tag{2.22}$$

$$\Delta H^\circ_{298} = 56.39 - 343.4(1-x) + 144.4(1-x)^2 \pm 6.0, \tag{2.23}$$

$$H^\circ_T - H^\circ_{298} = 41.40T + \left(4.678 \times 10^{-3}\right)T^2 + 25034x - 31.74xT - 4.150x^2T - 14488, \tag{2.24}$$

$$S^\circ_T = 95.34 \lg T + \left(9.356 \times 10^{-3}\right)T - 73.10x \lg T - 9.56x^2 \lg T + 185.07x + 24.19x^2 - 195.38, \tag{2.25}$$

where T is temperature, K, and x is the value of index in TaC_{1-x} formula [Eqs. (2.24–2.25) were recommended for the range of temperatures from 1200 to 2300 K]. The thermodynamic functions for stoichiometric tantalum monocarbide are tabulated by Turchanin et al. [52, 53] in the range of temperatures from 0 to 3000 K, by Schick [81] and Chase [579] in the range of 0–6000 K, by Barin [38] in the range of 298.15–3500 K, thermodynamic functions of $TaC_{0.99}$ in the range of 298.15–3000 K are tabulated by Storms [17] and Toth [115]; for stoichiometric tantalum semicarbide the thermodynamic functions are tabulated by Barin [38] in the range of 298.15–3773 K and by Schick [81] in the range of 0–3773 K.

During the vaporization processes from the surface of tantalum carbides in a vacuum at high and ultra-high temperatures, the composition of the carbide phase

(C/Ta ratio) can change significantly. The resulting composition gradient cannot be accurately evaluated, but it was found to become worse as the temperature was increased [17]. The following equations were recommended for tantalum (P_{Ta}, Pa) and carbon (P_C, Pa) partial pressures over near-stoichiometric monocarbide phase for the range of about 2500–3000 K (2200–2700 °C) [16, 17]:

$$\lg P_{Ta} = -\left(4.777 \times 10^4\right)/T + 12.174, \tag{2.26}$$

$$\lg P_C = -\left(3.234 \times 10^4\right)/T + 10.041, \tag{2.27}$$

the estimation of tantalum partial pressures (P_{Ta}, Pa) and C/Ta atomic ratios in the gaseous phase in the Ta–C system carried out by Kulikov [576] led to such relationships as

for the conditions of $Ta_{2\pm x}C$–TaC_{1-x} phases equilibrium:

at 1000–3270 K (730–3000 °C)

$$\lg P_{Ta} = -\left(4.377 \times 10^4\right)/T + 12.701, \tag{2.28}$$

$$\lg(C/Ta) = \left(2.087 \times 10^3\right)/T - 0.126, \tag{2.29}$$

at 3270–4270 K (3000–4000 °C)

$$\lg P_{Ta} = -\left(4.392 \times 10^4\right)/T + 12.748, \tag{2.30}$$

$$\lg(C/Ta) = \left(5.322 \times 10^3\right)/T - 1.115, \tag{2.31}$$

for the conditions of TaC_{1-x}–C phases equilibrium:

at 1000–3270 K (930–3000 °C)

$$\lg P_{Ta} = -\left(4.798 \times 10^4\right)/T + 12.087, \tag{2.32}$$

$$\lg(C/Ta) = \left(8.410 \times 10^3\right)/T + 3.191; \tag{2.33}$$

at 3270–4270 K (3000–4000 °C)

$$\lg P_{Ta} = -\left(4.684 \times 10^4\right)/T + 11.739, \tag{2.34}$$

$$\lg(C/Ta) = \left(6.157 \times 10^3\right)/T + 3.884, \tag{2.35}$$

at 4270–5700 K (4000–5400 °C)

$$\lg P_{Ta} = -\left(3.934 \times 10^4\right)/T + 9.925, \tag{2.36}$$

$$\lg(C/Ta) = \left(1.893 \times 10^4\right)/T + 1.105, \tag{2.37}$$

where T is temperature, K (*see* Table 2.5). The partial pressures of tantalum and carbon over the carbide phase as functions of composition at the fixed temperature 3000 K (2730 °C) are shown in Fig. 2.5 in comparison with the equilibrium pressure of carbon over graphite surface. At lower temperatures and short-term heating mainly carbon vaporized from the carbide surface in vacuum [16]. The solid monocarbide TaC_{1-x} loses carbon rapidly, being enriched with tantalum. Fesenko and Bolgar [125, 657] proposed for the evaporation rate of $TaC_{0.71}$ G, g cm^{-2} s^{-1}, as a function of temperature T, K, at 2970–3370 K (2700–3100 °C) the following equation:

$$\lg G = 9.10 - (4.49 \times 10^4)/T, \tag{2.38}$$

giving for 2700 and 3100 °C the values of 1.25×10^{-6} and 1.22×10^{-4} g cm^{-2} s^{-1}, while for $TaC_{0.93}$ Deadmore [327] measured at 2225 and 2625 °C the values of 5×10^{-8} and ~10^{-6} g cm^{-2} s^{-1}, respectively; and for $TaC_{0.96}$ (content O – 0.24%) Gusev [761] measured in vacuum (1.3×10^{-3} Pa) at 1900, 2200 and 2400 °C – 3.0×10^{-9}, 1.4×10^{-8} and 4.0×10^{-8} g cm^{-2} s^{-1}, respectively (corresponding calculated partial pressures of Ta – 2.2×10^{-5}, 1.1×10^{-4}, 3.4×10^{-4} Pa and C – 2.1×10^{-5}, 1.0×10^{-4}, 2.5×10^{-4} Pa, respectively), and established for the general evaporation rate G, g cm^{-2} s^{-1}, at 2100–2700 K (1830–2430 °C) the following relationship

$$\lg G = -(2.62 \pm 0.02) - (1.28 \times 10^4)/T. \tag{2.39}$$

Subsequently the monocarbide phase converts to two-phase (TaC_{1-x} + β-$Ta_{2\pm x}C$) area and then to the semicarbide phase area, where the congruently vaporizing composition shifts from $TaC_{0.53}$ at 2820 K (2550 °C) to $TaC_{0.47}$ at 3350 K (3080 °C) [35]. The small samples of monocarbide achieve the congruently vaporizing composition in 10 h at 2820 K (2550 °C) [65] and 100 h at 2400 K (2130 °C) [35]; in the range of temperatures from 2600 K (2330 °C) to 3400 K (3130 °C) this composition is described by the equation as follows [35, 65]:

$$\lg x = -6.8 \times 10^{-3} - \left(9.05 \times 10^2\right)/T, \tag{2.40}$$

where x is the value of index in TaC_{1-x} formula and T is temperature, K, and for the rate of congruent evaporation V, g cm^{-2} s^{-1}, in the Ta–C system at 3000–3400 K (2730–3130 °C) the equation is given as

$$\lg V = 8.725 - \left(4.2854 \times 10^4\right)/T, \tag{2.41}$$

while the similar equation was presented as well for the evaporation rate of near-stoichiometric β-$Ta_{2\pm x}C$ G, g cm^{-2} s^{-1}, as a function of temperature T, K, for the range of 3015–3355 K (2740–3080 °C) [1, 325]:

$$\lg G = 8.73 - \left(4.2854 \times 10^4\right)/T. \tag{2.42}$$

Table 2.5 Parameters of the gaseous phase in the Ta–C system in the conditions of $Ta_{2\pm x}C$–TaC_{1-x} and TaC_{1-x}–C phases equilibria calculated on the basis of thermodynamic data [576]

Parameters	Temperature, K (°C)									
	1000 (730)	2000 (1730)	2500 (2230)	3000 (2730)	3270 (3000)	3770 (3500)	4000 (3730)	4270 (4000)	5000 (4730)	5700 (5430)
	$Ta_{2\pm x}C$–TaC_{1-x} phases equilibrium									
$\lg P_{Ta}$, Pa	−31.068	−9.240	−4.851	−1.914	−0.684	1.101	1.804	2.379	3.798	4.826
$\lg P_{\Sigma}^{a}$, Pa	−29.102	−8.096	−3.941	−1.189	−0.060	1.606	2.203	2.855	4.029	4.950
C/Ta[b]	91.44	12.95	7.14	4.34	3.25	2.24	1.55	2.08	0.80	0.36
Contents, vol.%:										
C	98.92	92.81	87.53	80.57	75.32	67.76	58.40	63.75	40.92	24.18
C_2	−	0.02	0.15	0.53	0.83	0.78	1.48	2.44	2.10	1.14
C_3	−	−	0.01	0.05	0.09	0.20	0.18	0.35	0.09	0.02
Ta	1.08	7.17	12.31	18.85	23.76	31.26	39.94	33.46	56.83	74.66
	TaC_{1-x}–C phases equilibrium									
$\lg P_{Ta}$, Pa	−35.888	−11.848	−7.067	−3.898	−2.584	−0.692	−0.018	0.718	2.062	3.023
lg(C/Ta)	11.601	6.914	6.303	5.924	5.763	5.738	5.537	5.325	4.798	4.426

[a]Total gas pressure
[b]Atomic ratio in gaseous phase

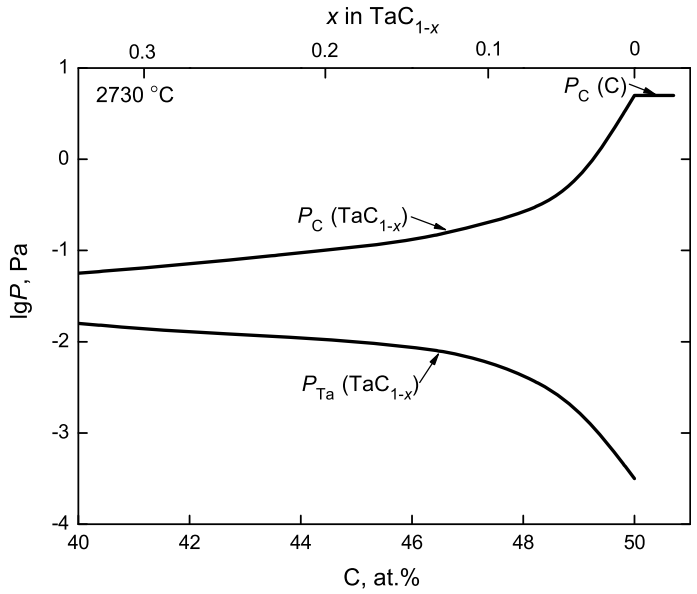

Fig. 2.5 Partial pressures of tantalum P_{Ta} (TaC_{1-x}) and carbon P_C (TaC_{1-x}) over tantalum monocarbide TaC_{1-x} phase, and vapour pressure of carbon over pure graphite P_C (C) as functions of monocarbide composition at 3000 K (2730 °C) calculated by Kaufman [16, 35, 115, 123, 363]

The values of general thermodynamic properties, vapour pressures and mass/linear vaporization rates for tantalum carbides are given in Addendum in comparison with other ultra-high temperature materials in the wide ranges of temperatures.

At room temperature thermal conductivity of near-stoichiometric tantalum monocarbide TaC_{1-x}, affected by porosity noticeably, is about 15–35 W m^{-1} K^{-1} (thermal diffusivity – 0.08 cm^2 s^{-1}), and for near-stoichiometric tantalum semicarbide α-$Ta_{2+x}C$ the corresponding value is about 35 W m^{-1} K^{-1} [35, 37, 41, 43, 66, 67, 115, 139, 544–546, 701, 721, 729, 771]. Within the homogeneity range, as a consequence of conduction electrons scattering on the carbon sublattice vacancies and thermal lattice vibrations, the thermal conductivity of TaC_{1-x} declines with increasing carbon deficit (value of x) in the phase [16, 35, 41] that is shown in Fig. 2.6. The variation of thermal conductivity with temperature for near-stoichiometric TaC_{1-x} on the basis of several sources is shown in Fig. 2.7.

The mean coefficients of linear thermal expansion of near-stoichiometric phases α-$Ta_{2+x}C$ and TaC_{1-x} at room temperature are (3.7–7.8) × 10^{-6} K^{-1} [35, 37, 97, 123] and (5.6–8.3) × 10^{-6} K^{-1} [46, 60, 72, 79, 93, 215, 582], respectively. The experimental data collected from the measurements of thermal expansion of

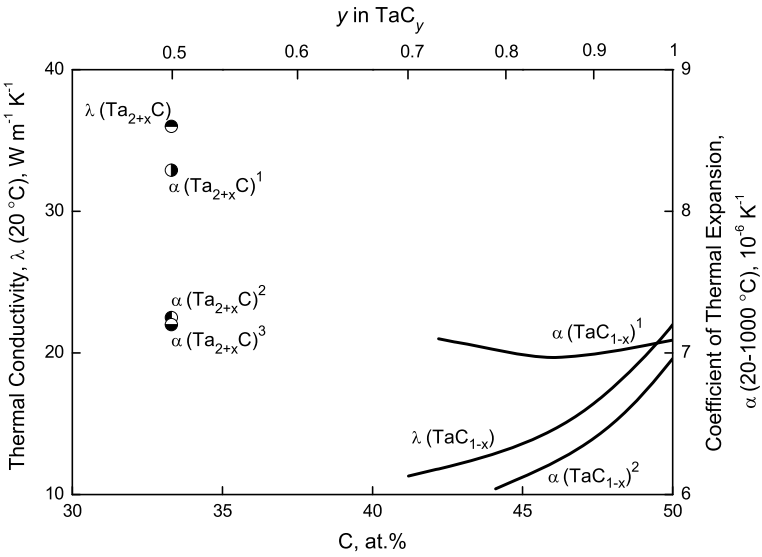

Fig. 2.6 Coefficients of thermal conductivity (λ) at 20 °C [35, 37, 41, 43] and linear thermal expansion (α) in the temperature range of 20–1000 °C (when it is not indicated specially) of tantalum monocarbide TaC_{1-x} (within the homogeneity range: 1 – hot-pressed and annealed, 6–10% porosity, content non-combined C ≤ 0.12% [37, 326, 582, 651, 666]; 2 – [48]) and semicarbide α-$Ta_{2+x}C$ (1 – in the range of 20–1200 °C [43]; 2 – hot-pressed and annealed, 6–10% porosity [582, 651]; 3 – in the range of 300–1200 °C [45, 828]) as functions of carbides composition

near-stoichiometric tantalum monocarbide TaC_{1-x} are listed in Table 2.6. The approximation function for the temperature dependence of relative thermal linear expansion $\Delta l/l_0 = f(T, \text{K})$, %, of near-stoichiometric TaC_{1-x} was recommended by Touloukian et al. [582] and based on the data for pure samples from several sources (accuracy within ±5% at $T < 2000$ K and within ±7% at $T > 2000$ K):

$$\Delta l/l_0 = -0.162 + (5.324 \times 10^{-4})T + (6.990 \times 10^{-8})T^2 + (2.780 \times 10^{-12})T^3, \tag{2.43}$$

where T is temperature, K, for 293–3200 K (20–2930 °C). On the basis of high-temperature X-ray measurements Houska [93] proposed for the average coefficient of linear thermal expansion α_m, K^{-1}, of near-stoichiometric sintered (in vacuum) $TaC_{\sim 1.0}$ (content O – 0.02%) in the range of 25–2040 °C the following equation:

$$\alpha_m = 6.00 \times 10^{-6} + (0.56 \times 10^{-9})(t - 25), \tag{2.44}$$

where t is temperature, °C. The influence of carbon content on the thermal expansion of monocarbide phase within the homogeneity range can be seen in Figs. 2.3 and 2.6. The average values of coefficient of linear thermal expansion of near-stoichiometric tantalum semicarbide α-$Ta_{2+x}C$ were evaluated as 7.2×10^{-6} K^{-1}

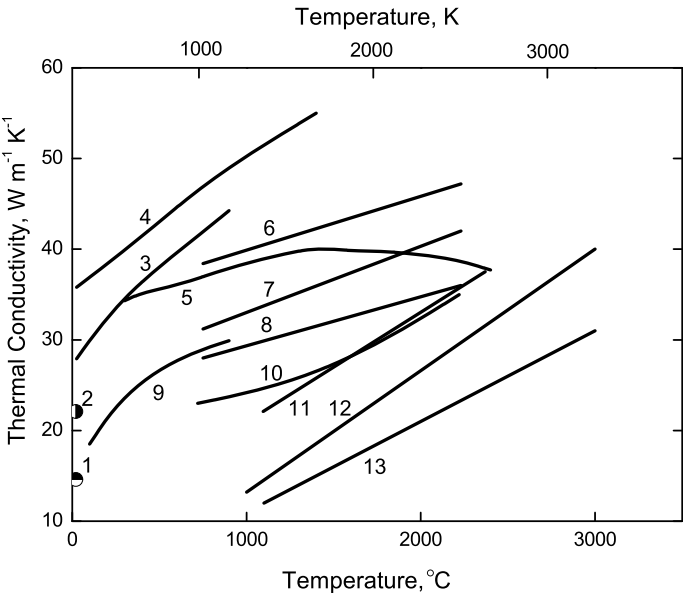

Fig. 2.7 Variation of thermal conductivity with temperature for tantalum monocarbide TaC_{1-x} materials on the basis of several sources: 1 – [66], 2 – [37, 67, 115, 545, 546], 3 – spark-plasma sintered, porosity – 1% [721], 4 – spark-plasma sintered, porosity < 2% [729], 5 – as-hot-pressed [51, 68], 6 – $TaC_{0.98}$ [585], 7 – $TaC_{0.80}$ [585], 8 – $TaC_{0.74}$ [585], 9 – spark-plasma sintered, porosity – ~7% [771], 10 – hot-pressed [337], 11 – hot-pressed and heat-soaked at 2480 °C [68], 12 – [69], 13 – hot-pressed [35, 123] (when it is not indicated specially, data are given for near-stoichiometric compositions)

in the range of temperatures from 300 to 1200 °C [45] and 8.3×10^{-6} K^{-1} in the range of 20–1200 °C [43]. On the basis of X-ray measurements Lönnberg [97] proposed for the average coefficients of thermal expansion α_m, K^{-1}, along the main crystallographic directions *a* and *c* of near-stoichiometric α-$Ta_{2+x}C$ in the temperature range of 25–1030 °C the following equations:

$$\alpha_{a,m} = 3.393 \times 10^{-6} + (2.834 \times 10^{-9})(t + 273), \tag{2.45}$$

$$\alpha_{c,m} = 1.769 \times 10^{-6} + (3.036 \times 10^{-9})(t + 273), \tag{2.46}$$

where $\alpha_{i,m}$ is the average coefficient of linear thermal expansion along the *i*-th direction, K^{-1}, and *t* is temperature, °C. In contrast to some other semicarbides of transition metals tantalum semicarbide α-$Ta_{2+x}C$ displays smaller thermal expansion coefficients perpendicular to the basal plane than parallel to it. The carbon atoms in the structure of α-$Ta_{2+x}C$ are more closely packed within the basal plane than perpendicular to it. It is most probable that this closer packing of the carbon atoms is responsible for the relatively high thermal expansion within the basal plane. Some recents works [204, 209, 413, 414, 561] are devoted to the calculations

Table 2.6 Average coefficients of linear thermal expansion α_m of near-stoichiometric tantalum monocarbide TaC_{1-x} in various temperature ranges

Temperature range, °C	α_m, 10^{-6} K^{-1}	References
20–200	5.89[a]	[582]
20–300	5.97[a]	[582]
20–400	6.03[a]	[582]
20–500	6.10[a]	[582]
	6.29 ± 0.29[b]	[79]
	6.29	[86, 545]
	6.32	[88]
	6.65	[61]
20–600	6.18[a]	[582]
	6.38	[43]
	6.41[c]	[87]
	6.45	[51]
20–700	6.25[a]	[582]
	6.45	[43]
	6.52[c]	[87]
	6.55	[43, 51]
20–730	6.31	[677, 678]
20–800	6.33[a]	[582]
	6.52	[43]
	6.56[c]	[87]
	6.60	[51]
	8.2	[89]
20–850	6.84	[61]
20–900	6.41[a]	[582]
20–1000	5.75[d]	[85]
	6.46 ± 0.14[b]	[79]
	6.49[a]	[582]
	6.50	[67]
	6.61[e]	[96, 546]
	6.64[c]	[87]
	6.65	[51]
	6.67	[88]
	6.96[f]	[48]
	7.1	[1, 37, 552]
	7.54	[41]
400–1000	4.09	[86]
20–1130	6.60	[677, 678]
20–1200	6.15[d]	[85]
	6.64[a]	[582]
	6.67[c]	[87]
	6.70	[51]
	6.84	[43]
	6.90	[61]
	7.13[g]	[93, 392]
	8.29	[72, 90]

(continued)

Table 2.6 (continued)

Temperature range, °C	α_m, 10^{-6} K^{-1}	References
20–1400	6.3[g]	[115, 392, 393, 550]
	6.4[g]	[68]
	6.40[d]	[85]
	6.80[a]	[582]
	6.89[c]	[87]
	6.90	[51]
	7.02	[43]
20–1500	6.52[d]	[85]
	6.67[e]	[96]
	6.88[a]	[582]
	7.05	[43, 51]
	7.11[c]	[43, 87]
	7.12	[88]
	8.2	[94]
20–1600	6.58[d]	[85]
	6.96[a]	[582]
	7.15	[43]
	7.25	[51]
	7.29[c]	[87]
20–1730	6.96	[16]
	7.08	[677, 678]
20–1800	6.66[d]	[85]
	7.12[a]	[582]
	7.34	[43]
	7.56	[51]
	7.59[c]	[87]
20–2000	6.88[d]	[85, 392]
	7.08 ± 0.33[h]	[729]
	7.29[a]	[582]
	7.31[e]	[96, 701]
	7.64	[88]
	7.90	[51]
	7.94[c]	[87]
20–2050	5.0[i]	[68, 92]
20–2130	7.41	[677, 678]
20–2200	7.46[a]	[582]
	9.09	[61]
20–2300	6.78[d]	[85]
	8.08	[43]
	8.20[c]	[87]
	8.30	[43, 51]
20–2400	7.62[e]	[96]
	7.63[a]	[582]
20–2450	5.5[i]	[68, 92]

(continued)

Table 2.6 (continued)

Temperature range, °C	α_m, 10^{-6} K^{-1}	References
20–2500	6.81[d]	[85]
	7.50	[43, 51]
	7.71[a]	[582]
	8.40	[88]
	8.46[c]	[43, 87]
20–2600	7.80[a, e]	[96, 582, 701]
	8.61[c]	[51, 87]
20–2650	8.02	[86]
	8.2 ± 0.8	[43, 67, 91, 581]
20–2800	7.97[a]	[582]
20–2900	8.06[a]	[582]
	8.81[c]	[87]
	8.85	[51]
20–3000	8.4[j]	[95]

[a]Calculated on the basis of approximation function
[b]$TaC_{0.98}$, measured by high-temperature X-ray diffraction
[c]Hot-pressed $TaC_{0.99}$ (27% porosity)
[d]Hot-pressed $TaC_{0.98}$ (13% porosity), measured by an optical-micrometer method in Ar atmosphere
[e]$TaC_{0.99}$ hot-pressed at 3100–3150 °C and measured by dilatometric method
[f]$TaC_{0.997}$ measured by high-temperature X-ray diffraction
[g]Measured by high-temperature X-ray diffraction
[h]Spark-plasma sintered (porosity <2%, mean grain size – 5.8 ± 0.2 μm), measured by dilatometric method in Ar atmosphere
[i]Hot-pressed $TaC_{0.99}$ (25% porosity)
[j]Hot-pressed $TaC_{0.97}$ measured by a dilatometric method

of coefficients of thermal expansion of tantalum carbides by means of theoretical modelling.

In comparison with other ultra-high temperature materials the values of thermal conductivity and thermal expansion of tantalum carbides in the wide range of temperatures are summarized in Addendum.

2.3 Electro-magnetic and Optical Properties

At room temperature the value of specific electrical resistance (resistivity) of near-stoichiometric tantalum monocarbide TaC_{1-x} lies within the area of 0.2–0.5 μΩ m, and for near-stoichiometric semicarbide α-$Ta_{2+x}C$ it is about 0.4–0.5 μΩ m [1, 16, 35, 43, 60, 61, 90, 207, 372, 382, 388, 534, 544, 676, 830]. The variation of specific electrical resistance with temperature for near-stoichiometric TaC_{1-x} is shown on the basis of several sources in Fig. 2.8. In the wide temperature range from 300–500 up to 3000–3300 °C the resistance of monocarbide phase enlarges with increasing temperature, practically in accordance with linear relationship; that

is an evidence of mainly metallic type of conduction in TaC_{1-x} [35]. From the room temperature up to 800 °C, according to Nino et al. [721], the electrical resistivity of $TaC_{\sim 1.0}$ materials (porosity – 1%) ρ, μΩ m, is obeyed the rule:

$$\rho = 0.35 + (5.70 \times 10^{-4})t, \tag{2.47}$$

for the interval of 1130–3230 °C Eckstein and Forman [102] recommended for near-stoichiometric tantalum monocarbide TaC_{1-x} the equation as follows:

$$\rho = 1.02 + (1.78 \times 10^{-4})(t + 273), \tag{2.48}$$

where t is temperature, °C. The single crystal $TaC_{0.99}$ resistivity measured by Allison et al. [100] at ultra-low and low temperatures is perfectly described in the interval from T = 0 to T = 350 K by the Wilson expression (in μΩ m):

$$\rho(T) = 0.125 + (5.353 \times 10^{-8})T^3 \int_0^{136/T} x^3 / \sinh^2 x dx, \tag{2.49}$$

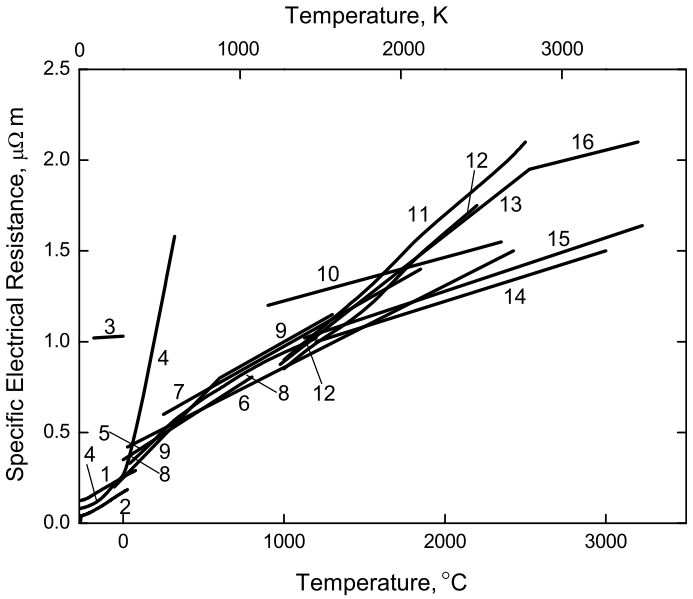

Fig. 2.8 Variation of specific electrical resistance with temperature for tantalum monocarbide TaC_{1-x} materials on the basis of several sources: 1 – single crystal $TaC_{0.99}$ [100]; 2 – hot-pressed $TaC_{0.997}$, porosity – ~11% [99]; 3 – sintered $TaC_{\sim 1.0}$, porosity – 13% [104]; 4 – [51]; 5 – [103]; 6 – [105]; 7 – [98]; 8 – spark-plasma sintered, porosity – 1% [721]; 9 – sintered (in vacuum) $TaC_{0.98}$ [584, 585]; 10 – [345]; 11-12 – [61]; 13 – [101]; 14 – [69]; 15 – metal carburized in hydrocarbon atmosphere [102]; 16 – [106] (when it is not indicated specially, data are given for near-stoichiometric compositions)

Table 2.7 Average thermal coefficients of resistivity α_R of near-stoichiometric tantalum monocarbide TaC_{1-x} in various temperature ranges

Temperature range, °C	α_R, 10^{-3} K^{-1}	References
(–225)–(+20)	4.0[a]	[100]
(–195)–0	0.53[b]	[365]
(–190)–(+20)	0.25	[366]
20–25	1.5–2.5[c]	[390]
	2.5[d]	[206]
0–800	1.63[e]	[721]
20–1000	0.79[f]	[371, 583]
100–1100	1.38	[368, 369, 648]
20–1170	0.22	[367, 369, 648]
400–2000	1.07[g]	[369, 648]
	0.99	[370]
20–2110	0.22	[367, 369, 648]

[a]Single crystal $TaC_{0.99}$, $d\rho/dT = 4.95 \times 10^{-4}$ μΩ m K^{-1}
[b]Materials deposited from gas phase
[c]Coating reactively sputter-deposited from Ar plasmas containing CH_4, thickness – 1–25 μm
[d]Hot-pressed materials, carbon content – 49.5 at.%
[e]Spark-plasma sintered materials (porosity – 1%)
[f]Thermal coefficient $\alpha = 1/\rho_\theta \times (d\rho/dT)_\theta$, corresponding to Debye temperature $\theta = 630$ K ($(d\rho/dT)_\theta = 3.3 \times 10^{-4}$ μΩ m K^{-1}), corrected to the poreless state
[g]Hot-pressed materials

which can be simplified for $T \leq 25$–40 K up to $\rho(T) = 0.125 + AT^3$, where A is a constant. The data on the thermal coefficients of resistivity of near-stoichiometric TaC_{1-x} in the various temperature ranges are presented in Table 2.7. The variations of resistivity with temperature for the phases having different deviations from the stoichiometry are presented in Fig. 2.9. The specific electrical resistances in solid and liquid states and ratio of the resistivities at the melting point of TaC_{1-x} are $\rho_{sol} \approx 2.2$–2.3 μΩ m, $\rho_{liq} \approx 2.4$ μΩ m and $\rho_{liq}/\rho_{sol} \approx 1.04$–$1.09$, respectively [112, 113].

Within the homogeneity range the resistivity of TaC_{1-x} enlarges considerably with increasing carbon deficit in non-metal sublattice, that is shown in Fig. 2.10, where the data from several sources are collected [1, 23, 37, 103, 107–111]; the variation of $d\rho/dT$ with composition at ambient temperatures is shown in Fig. 2.11. The superconducting transition temperature T_c of near-stoichiometric TaC_{1-x} is 9.1–11.0 K [35, 60, 73, 114–116, 356–358, 451, 661, 665, 688, 832]; it varies with composition considerably, e.g. falling from 9.7 K for $TaC_{0.99\pm0.01}$ up to 2.0 K for $TaC_{0.85\pm0.01}$ [114]. The highest temperature is reached for the most stoichiometric composition and any deviation therefrom results in a lowering and eventual disappearance (<1.05 K at $x \geq 0.25$–0.30) of T_c [114–116]. The transition temperature T_c of semicarbide α-$Ta_{2+x}C$ phase having $x = 0.13$ is <1.60–1.75 K [115, 116], for the phase with $x = 0.08$ it is <1.98 K [831, 832], and for the stoichiometric phase $T_c = 3.26$ K [586].

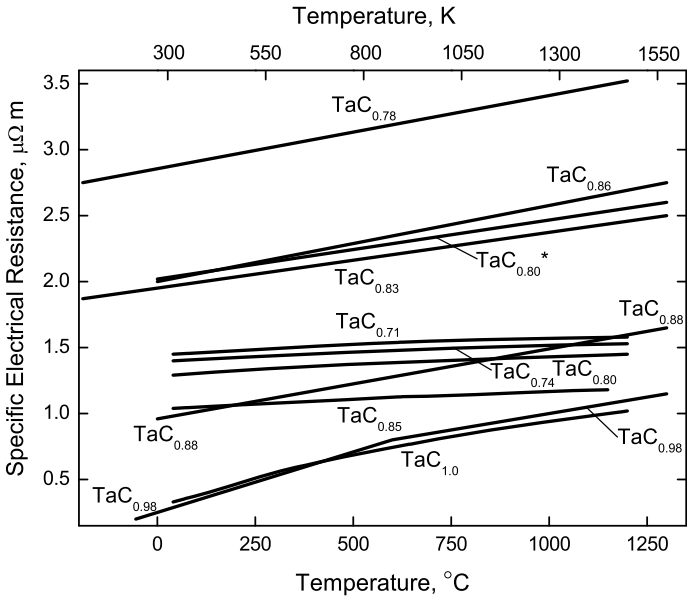

Fig. 2.9 Variations of specific electrical resistance with temperature for the TaC_{1-x} phases having different deviations from the stoichiometry (x = 0.0, 0.15, 0.20, 0.26, 0.29 [103]; sintered materials, x = 0.02, 0.12, 0.14, 0.17, 0.20*, 0.22 [584, 585])

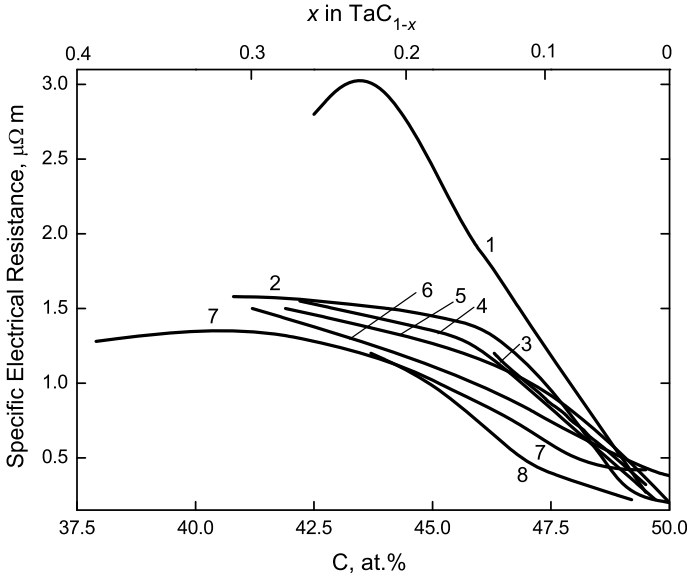

Fig. 2.10 Variations of specific electrical resistance at room temperature within the homogeneity range of tantalum monocarbide TaC_{1-x} materials on the basis of several sources: 1 – materials sintered in vacuum [584, 585], 2 – materials carburized by solid-state diffusion saturation [107], 3 – [321], 4 – [1, 23, 37, 111], 5 – materials carburized in propane [322, 522], 6 – [103, 108], 7 – [109, 660], 8 – [110]

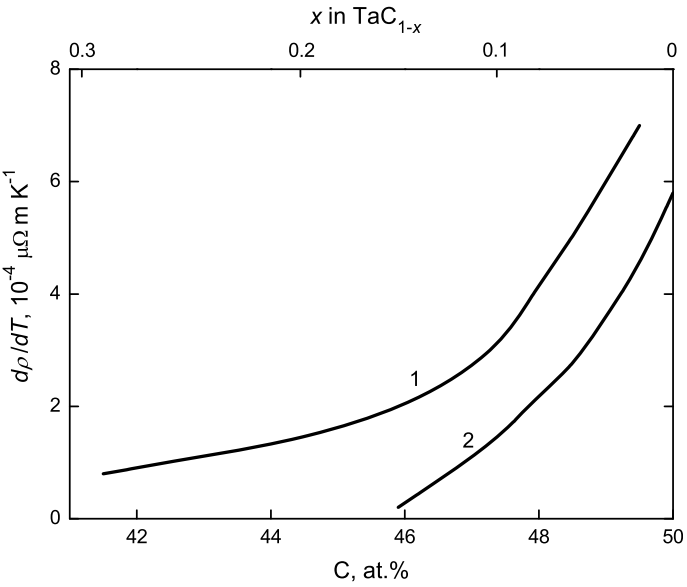

Fig. 2.11 Variation of $d\rho/dT$ (slope of resistivity vs. temperature) at ambient temperatures with carbon content within the homogeneity range of tantalum monocarbide TaC_{1-x} based on: 1 – [103] and 2 – [107]

At room temperature the Hall and Seebeck coefficients of near-stoichiometric tantalum monocarbide TaC_{1-x} are $R = -(0.6–2.1) \times 10^{-10}$ m^3 A^{-1} s^{-1} and $S = -(4.0–11.5)$ μV K^{-1}, respectively [35, 37, 60, 111, 117, 328, 585, 587]; the reported values of these constants for nonstoichiometric TaC_{1-x} phases show considerable disagreement [35, 37, 108, 111, 123], experimental data obtained by Santoro and Dolloff [322, 522] for the Hall coefficient, Neshpor and Ordanyan [108] for the Seebeck coefficient and Borukhovich et al. [111, 585] for the Hall and Seebeck coefficients within the homogeneity range of tantalum monocarbide TaC_{1-x} for ambient and cryogenic temperatures are presented in Fig. 2.12.

Near-stoichiometric tantalum monocarbide TaC_{1-x} is a paramagnetic substance with molar magnetic susceptibility χ_m (SI) ≈ (120–250) × 10^{-6} cm^3 mol^{-1} at room temperature very slightly varying with temperature in the range from −200 to 900 °C [35, 107, 118–120, 328, 329, 662]. The weak temperature influence and small absolute value of susceptibility indicate that the magnetic properties of TaC_{1-x} are determined primarily by conducting electrons [123]. With increasing carbon deficit within the homogeneity range the value of magnetic susceptibility of TaC_{1-x} declines to its diamagnetic (negative) minimum at −(130–320) × 10^{-6} cm^3 mol^{-1}, corresponding to ~$TaC_{0.80}$ composition (*see* Fig. 2.13), and then it enlarges again proceeding closer to the paramagnetic region [35, 118, 119, 122, 373, 589]. The temperature coefficient of magnetic susceptibility $d\chi/dT$ changes its sign with composition; it is positive for compositions near $TaC_{0.78}$ and negative

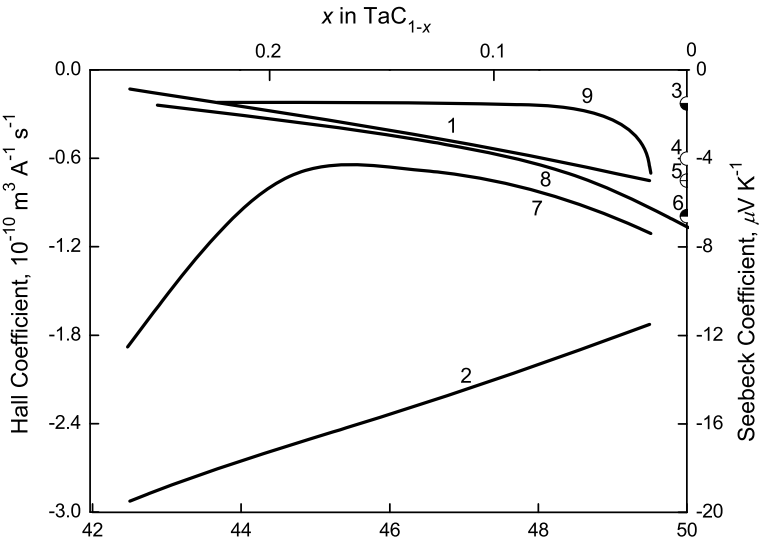

Fig. 2.12 Variations of Seebeck coefficient at room temperature (1 – [108]; 2 – [111, 585]; 3 – [588]; 4 – thin films deposited by r. f. sputtering technique, thickness – 0.3 μm [117]; 5 – hot-pressed materials [105]; 6 – [328]) and Hall coefficient at room (7 – sintered in vacuum [111, 585]; 8 – materials carburized in propane [322, 522]) and cryogenic, 77 K (9 – carburized in propane [322, 522]) temperatures within the homogeneity range of tantalum monocarbide TaC_{1-x}

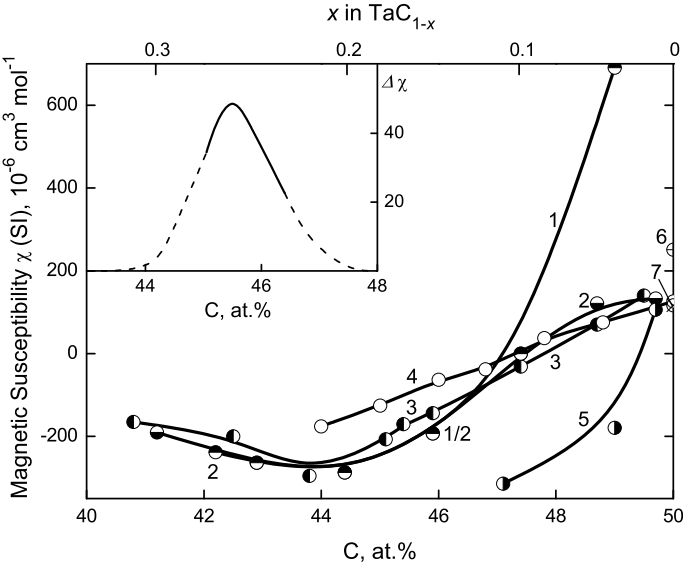

Fig. 2.13 Variation of molar magnetic susceptibility at room temperature within the homogeneity range of tantalum monocarbide TaC_{1-x} on the basis of several sources: 1 – [110]; 2 – contents: non-combined C ≤ 0.005%, N ≤ 0.005%, O ≤ 0.02% [122]; 3 – disordered (for ordered state *see* data in the *Inset*) [119, 121, 654]; 4 – from 48.8 to 50 at.% C the values extrapolated [118, 589]; 5 – [107]; 6 – [329]; 7 – [328] (*Inset* – difference of the susceptibility magnitudes in disordered and ordered states $\Delta\chi = |\chi_{disord} - \chi_{ord}|$ determined for $TaC_{0.82}$–$TaC_{0.85}$ phases, units are the same with the major plot [121, 654])

for – near-stoichiometric compositions around $TaC_{0.97}$ [35, 119]. The χ–T curves of the monocarbide TaC_{1-x} with $x \leq 0.1$ and $x \geq 0.2$ exhibit no peculiarities since these compositions lie outside the $TaC_{0.81}$ to $TaC_{0.89}$ range, in which due to the ordering of carbon atoms and vacancies in TaC_{1-x} the reversible susceptibility jump corresponding to the region of temperature hysteresis is observed [121]. According to Santoro [122] tantalum semicarbide α-$Ta_{2+x}C$ is also a paramagnetic substance with the value of its mass magnetic susceptibility something higher than that of near-stoichiometric monocarbide TaC_{1-x}, but ζ-Ta_4C_{3-x} is more likely diamagnetic.

According to Harris et al. [126] tantalum monocarbide TaC_{1-x} films (thickness 0.2–0.3 μm) in the condition of as-grown on metallic tantalum single-crystal with (100) orientation exhibit reflectance (reflective index) – 0.3-0.4 at a wavelength of about 0.1 μm. The variation of reflectance with angle of incidence for TaC_{1-x} film in high-vacuum ultraviolet range is given in Fig. 2.14, and the room-temperature reflectance spectrum of pure sintered near-stoichiometric TaC_{1-x} (porosity – 10%, grain size – 10-20 μm) from the ultraviolet wavelength region to the mid-infrared band is shown in Fig. 2.15. Optical properties of a nearly stoichiometric single crystal of tantalum monocarbide (specular reflectance at 0.03–6.0 eV and ellipsometry at 1.2–4.6 eV) were reported by Modine et al. [124], spin-resolved photoemission spectra of $TaC_{0.96}$ – by Garbe and Kirschner [671] and X-ray photoelectron spectra of TaC_{1-x} phases within the homogeneity range ($0 < x \leq 0.48$) – by Gruzalski and Zehner [670]. Optical spectra in the infrared (IR) and visible ranges as well as X-ray emission and absorption spectra were considered by Alyamovskii et al. [330], Ramqvist et al. [158], Kammori et al. [833] and

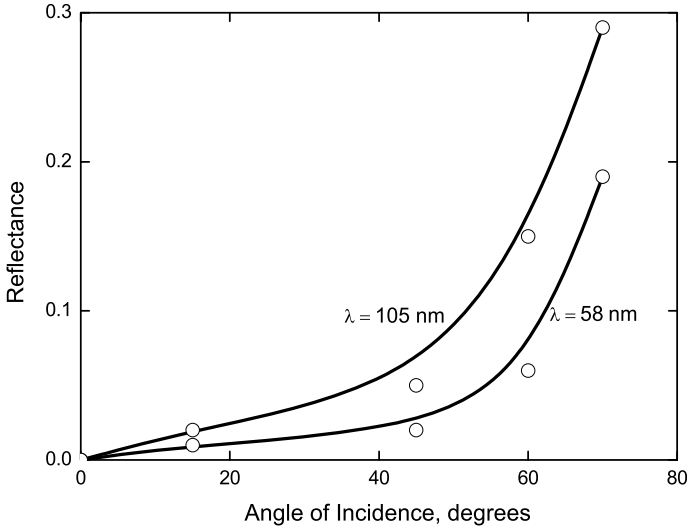

Fig. 2.14 Variation of reflectance with angle of incidence for as-grown (on metallic tantalum single-crystal with (100) orientation) TaC_{1-x} film (thickness – 0.2-0.3 μm) in high-vacuum ultraviolet range [126]

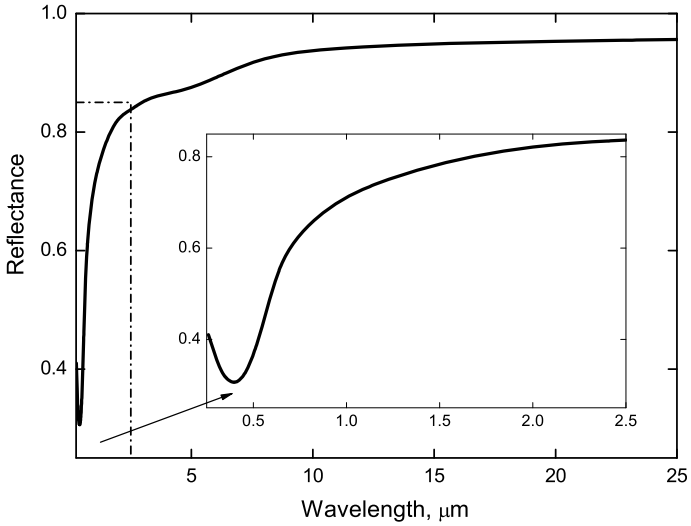

Fig. 2.15 Room-temperature reflectance spectrum of pure sintered TaC_{1-x} (porosity – 10%, grain size – 10-20 μm) from the ultraviolet wavelength region to the mid-infrared band [127, 128]

Upadhyaya [123]. The absorption wave numbers of basic maxima of IR band for tantalum carbides are 1080 cm^{-1} (for $TaC_{0.96}$), 1100 cm^{-1} (for $TaC_{0.73}$) and 1080 cm^{-1} (for $Ta_{1.96}C$) [330]. At the common conditions the colour of carbon-rich tantalum monocarbide TaC_{1-x} materials is brown in the dispersed state (powdered) or golden-brown (golden-yellow) in the compact state, the colour of tantalum semicarbide α-$Ta_{2+x}C$ materials is gray [17, 37, 43, 60]. The colour intensity of TaC_{1-x} compact materials decreases sharply with an increase in deviation from the stoichiometry, so TaC_{1-x} phases with the value of index $x > 0.15$ are steel or silver in colour, i.e. very similar to that of transition metal carbides of group 4 [123].

The normal monochromatic emittance (spectral emissivity) $\varepsilon_{\lambda,T}$ of tantalum monocarbide TaC_{1-x} falls noticeably with wavelength λ growth and only slightly varies with temperature [16, 61, 123, 539]; the linear relationships for λ = 0.4–1.1 μm

$$\varepsilon_{\lambda,T} = (0.693 \pm 0.015) - [(7.0 \pm 0.5) \times 10^{-5}](t + 273) - (0.147 \pm 0.01)\lambda, \tag{2.50}$$

where t is temperature, °C, λ is wavelength, μm, was proposed by Danilyants et al. [151] in the temperature range 1430–3130 °C for materials manufactured by remelting of pressed briquets with composition $TaC_{0.98}$, and for λ = 0.65 μm

$$\varepsilon_{\lambda} = 0.4662 - (5.084 \times 10^{-5})(t + 273), \tag{2.51}$$

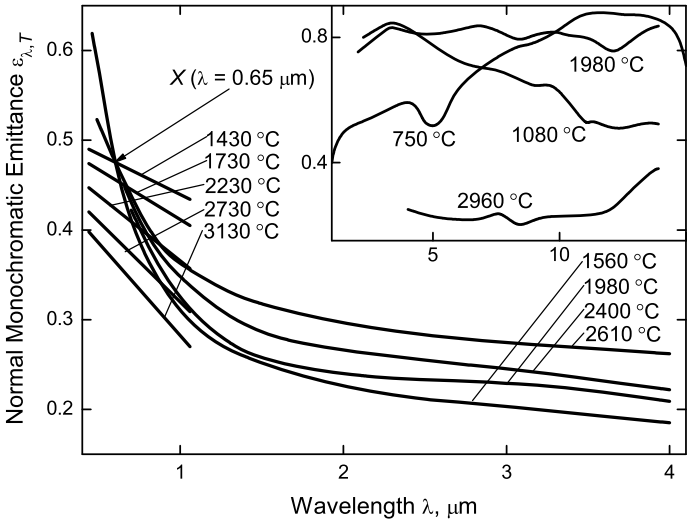

Fig. 2.16 Variations of normal monochromatic emittance $\varepsilon_{\lambda,T}$ with wavelength λ for tantalum monocarbide TaC_{1-x} materials on the basis of several sources: remelted $TaC_{0.98}$, measured in high-purity Ar at 1430, 1730, 2230, 2730 and 3130 °C [151]; hot-pressed $TaC_{\sim 1.0}$, measured in vacuum at 1560, 1980, 2400 and 2610 °C (*X*-point, or an isosbestic point, where $|1/\varepsilon_{\lambda,T} \times d\varepsilon_{\lambda,T}/dT|_{\lambda=X} = 0$, is marked specially) [138, 590, 595] (*Inset* – $TaC_{\sim 1.0}$, measured in Ar + H_2 atmosphere at 1080, 1980 and 2960 °C [596, 597]; plasma-sprayed coating $TaC_{\sim 1.0}$, measured at 750 °C [598, 599])

where *t* is temperature, °C, was obtained by Mackie et al. [129, 539] in the temperature range of 1100–1700 °C for arc-melted (via floating zone) polycrystalline monocarbide $TaC_{0.79\pm0.05}$. The variation of emittance $\varepsilon_{\lambda,T}$ with wavelength and temperature for tantalum monocarbide is shown in Fig. 2.16. The data on monochromatic emittance ε_λ (λ = 0.630–0.665 μm) and integral (total) emittance ε_T for TaC_{1-x} materials produced by different manufacturing methods and measured at various temperatures are listed in Tables 2.8 and 2.9. According to Bober et al. [152], tantalum monocarbide shows in the solid and liquid states only a small variation of emissivity at λ = 0.63 μm with temperature.

The thermionic emission characteristics (electron work function and Richardson constants) of various tantalum carbide phases are given in Fig. 2.17 and Table 2.10. The calculated values of emission current density for near-stoichiometric monocarbide TaC_{1-x} are about 0.2, 10^2 and 9×10^3 A m^{-2} at 730, 1230 and 1730 °C, respectively [16, 143]. Similar to some other transition metal carbides, single crystal TaC_{1-x} materials show following relationships between the effective thermionic work functions of different crystallographic planes: $\varphi_{(210,\ 310)} < \varphi_{(100)} \approx \varphi_{(110)} \gg \varphi_{(111)}$ [153]. A typical brightness characteristic (U = 75 kV) of tantalum monocarbide

Table 2.8 Normal monochromatic emittance ε_λ (λ = 0.630–0.665 μm) of tantalum monocarbide TaC_{1-x} materials in various temperature ranges

Monochromatic emittance, ε_λ	Temperature range, °C	References
0.36–0.47[a]	1430–3130	[151]
0.37–0.40[b]	1150–1730	[129, 539]
0.40–0.46[c]	2100–3200	[102]
0.43[d]	1500–2100	[374]
0.43–0.53[e]	1600–2700	[130]
0.45–0.47[f]	1920–2110	[131]
0.45[g]	1530–2530	[35, 138]
0.49–0.87[h]	1550–3600	[594]
0.50–0.52[i]	930–2830	[152]
0.50–0.85 (?)[j]	1500–3600	[135]
0.55[k]	2430–2730	[16, 35]
0.54–0.83[l]	1000–3000	[591, 833]
0.62–0.66[m]	1000–2200	[1, 132]
0.62–0.85[n]	800–1700	[37, 61, 72, 134]
0.67	1730	[16, 67]
0.80[i]	1500	[133]

[a]Remelted materials; ε_λ falls with the temperature growth linearly
[b]Arc-melted (via floating zone) polycrystalline $TaC_{0.79\pm0.05}$, measured in vacuum 10^{-5}–10^{-6} Pa; ε_λ falls with the temperature growth linearly (slope is ~5.1×10^{-5} K^{-1})
[c]Materials manufactured by metal carburization; ε_λ grows with temperature increasing
[d]For λ = 0.81 μm, polished surface (with rough surface ε_λ = 0.61), after exposure to hot H_2 at 1120 °C for 1 h ε_λ increases to a value similar to that of material with a rough surface [375]
[e]Hot-pressed $TaC_{0.93}$ (5–15% porosity); ε_λ grows with temperature increasing
[f]ε_λ grows with temperature increasing
[g]Sintered materials
[h]Specially polished materials; ε_λ falls with the temperature growth
[i]Hot-pressed materials; ε_λ grows very slightly with the temperature growth (linearly)
[j]Sharp decrease of emittance ε_λ was observed at ~2000 °C; however, careful studies of this effect in carbide materials [123, 136, 137] showed that it results from the formation and subsequent thermal desorption of surface impurity films formed by interactions with oxygen and nitrogen
[k]Materials manufactured by metal carburization
[l]Sintered materials, measured in vacuum and Ar multiply; ε_λ falls with the temperature growth linearly
[m]Sintered materials; ε_λ falls with temperature increasing
[n]Powdered materials on tantalum substrate; ε_λ grows with temperature increasing (almost linearly)

cathode (thermionic emitter for electron microscopy) is shown in Fig. 2.18. Field emission properties of TaC_{1-x} materials were reported in several works [153, 376, 377, 607, 608, 837].

Table 2.9 Integral (total) emittance ε_T of near-stoichiometric tantalum monocarbide TaC_{1-x} in various temperature ranges

Integral emittance, ε_T	Temperature range, °C	References
0.20–0.30[a]	1300–2200	[138, 590]
0.25–0.92[b]	500–2750	[592, 593]
0.36–0.44[c]	1600–2700	[139]
0.39[d]	2100–3200	[35, 102]
0.42–0.44[e]	1000–2700	[1, 140]
0.42–0.54[f]	1100–2900	[591, 834]
0.54–0.59[g]	2000–3000	[35]

[a]Normal emittance ε_T for hot-pressed materials, measured in Ar; ε_T increases with the temperature growth linearly

[b]Normal emittance ε_T for hot-pressed materials, measured in Ar; dome-shaped curve $\varepsilon_T - T$ with the maximum at ~1100 °C (ε_T = 0.37 at 500 °C)

[c]ε_T increases with the temperature growth

[d]Manufactured by metal carburization

[e]Normal emittance ε_T for hot-pressed $TaC_{0.98\pm0.03}$ (20–25% porosity); ε_T increases slightly with the temperature growth

[f]Hemispherical emittance ε_T for sintered materials; ε_T increases with the temperature growth

[g]Normal emittance ε_T for hot-pressed materials; ε_T increases slightly with the temperature growth

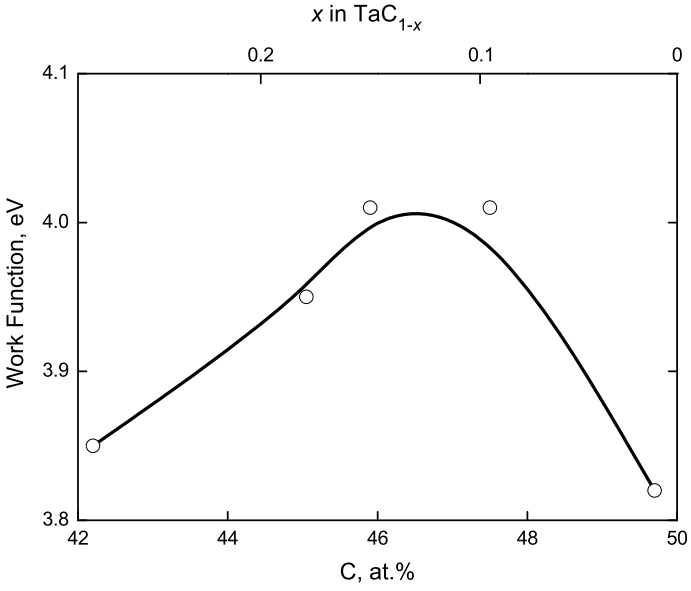

Fig. 2.17 Variation of effective work function of tantalum monocarbide phase TaC_{1-x} at 1530 °C with carbon content ("total current" method) [35, 123, 141, 658, 666]

Table 2.10 Thermionic emission characteristics (electron work function and Richardson constant) of tantalum carbide phases

Composition	Work function[a], $\varphi = \varphi_0 + (d\varphi/dT)_{av}T$, eV	Richardson constant, A, 10^4 A m^{-2} K^{-2}	Temperature range, °C	Remarks[b]	References
Ta_2C	4.11	–	–	LMTO-calculated value	[673, 743, 744]
	$4.31 + (1.0 \times 10^{-4})T$	–	700–1300	TiM, metal-carburized	[1]
	4.73[c]	–	–	UPS, single crystal (100)	[672–674, 743, 744]
$TaC_{0.73}$	3.85[d]	–	1500	TiM, hot-pressed	[35, 123, 604]
	$4.17 - (1.8 \times 10^{-4})T$	–	1250–1800	TiM, hot-pressed	[1, 35, 142, 604]
$TaC_{0.77}$	3.40[d]	–	1500	FEM, single crystal (100)	[153]
$TaC_{0.82}$	3.95[d]	–	1500	TiM, hot-pressed	[35, 123, 604]
	$4.31 - (2.0 \times 10^{-4})T$	–	1200–1800	TiM, hot-pressed	[1, 142]
$TaC_{0.85}$	4.01[d]	–	1500	TiM, hot-pressed	[35, 123, 604]
	4.30 ± 0.05	–	25	CPDM (W), hot-pressed	[1, 146]
	$4.46 - (2.5 \times 10^{-4})T$	–	1200–1800	TiM, hot-pressed	[1, 35, 142, 604]
$TaC_{\sim 0.90}$	3.80 ± 0.05[d]	–	–	TiM, metal-carburized	[1, 142]
	3.80 ± 0.1	165	–	TiM, metal-carburized	[1, 142]
$TaC_{0.91}$	4.01[d]	–	1500	TiM, hot-pressed	[35, 123, 604]
	$4.37 - (2.0 \times 10^{-4})T$	–	1200–1800	TiM, hot-pressed	[1, 35, 142, 604]
$TaC_{0.98}$	4.22 ± 0.06	–	–	CPDM, PeM, carburized	[1, 147]
$TaC_{0.99}$	3.82[d]	–	1500	TiM, hot-pressed	[35, 123, 604]
	$4.18 - (2.0 \times 10^{-4})T$	–	1200–1800	TiM, hot-pressed	[1, 35, 142, 604]
$TaC_{\sim 1.0}$	3.05	2.1	–	TiM, powder	[1, 143]
	3.14	0.30	–	TiM, outgased powder	[1, 143]
	3.17 ± 0.06	0.22 ± 0.10	1300–2000	TiM, metal-carburized	[1, 102]
	From (3.21 ± 0.05)[d] to (3.87 ± 0.1)[d]	–	From 1000 to 1300	TiM, metal-carburized (variation of φ_{eff} with temperature)	[1, 148]
	3.33[e]	–	–	Single crystal (001), theoretically evaluated	[417]
	3.40	–	–	TiM, pyrolytic coating	[1, 378]
	3.61[f]	–	–	Single crystal (001), theoretically evaluated	[417]

(continued)

Table 2.10 (continued)

Composition	Work function[a], $\varphi = \varphi_0 + (d\varphi/dT)_{av}T$, eV	Richardson constant, A, 10^4 A m^{-2} K^{-2}	Temperature range, °C	Remarks[b]	References
	3.64[d]	–	1900	TiM, powder on W substrate	[1, 143]
	3.65[d]	–	1700	TiM, powder	[145]
	3.77[d]	–	1100	TiM, powder	[1, 145]
	3.77[d]	–	2100	TiM, hot-pressed	[142]
	3.80 ± 0.05	–	1200–1700	TiM, metal-carburized	[679]
	3.80[d]	–	2100	TiM, powder on W substrate	[1, 142]
	3.84[g]	–	–	LMTO-calculated value for (001) surface	[673, 674, 743, 744]
	3.85	–	–	Calculated value (unrelaxed surface)	[743, 744]
	3.86	–	–	Theoretically evaluated (unrelaxed surface)	[673]
	3.93[d]	–	25	TiM, powder	[145]
	3.95[d]	7.64	2350	TiM, metal-carburized wire	[150]
	$3.98 - (1.5 \times 10^{-4})T$	–	1100–1900	TiM, powder	[1, 35, 123, 145]
	4.15[d]	–	1500	TiM, hot-pressed	[1, 144]
	4.16	–	–	Calculated value (relaxed surface)	[743, 744]
	4.2	–	–	Bulk materials for the CMOS device applications	[836]
	4.20 ± 0.04	–	1300	TiM, hot-pressed	[1, 149]
	4.24	–	–	Theoretically evaluated (relaxed surface)	[673]
	4.30[h]	–	25	FEM, single crystal (001), clean surface	[542, 607–609, 743, 744]
	4.33[d]	142.7	2350	TiM, electrolytic etching	[150]
	4.36[i]	–	–	Single crystal (100), theoretically evaluated	[684]
	4.38 ± 0.05	–	25	CPDM, hot-pressed	[1, 149]
	4.38[c]	–	–	UPS, single crystal (100)	[672–674, 743, 744]
	4.4[d]	–	2100	TiM, sintered	[1, 142]
	4.5 ± 0.1	–	–	TiM, metal-carburized	[1, 142]
	4.54[d]	4.01	2350	TiM, powder on Pt substrate	[150]
	4.70[j]	–	25	FEM, single crystal (111), clean surface	[542, 607–609]

(continued)

Table 2.10 (continued)

Composition	Work function[a], $\varphi = \varphi_0 + (d\varphi/dT)_{av}T$, eV	Richardson constant, A, 10^4 A m^{-2} K^{-2}	Temperature range, °C	Remarks[b]	References
	4.7	–	–	Nanocrystalline thin films sputtered on thermally oxidized Si wafers	[835]
	–	–	–	AeE, cold cathode, ~1 μm fibre, 800 A m^{-2} in vacuum 5.3 × 10^{-7} Pa for >1200 h	[1, 376, 377]

[a]T is temperature, K
[b]Methods applied for the experimental determination of the work function (TiM – thermionic method, UPS – ultraviolet photoemission spectroscopy technique, CPDM – contact potential difference method (second electrode is given in brackets), PeM – photoelectronic method, FEM – field emission microscopy, AeE – autoelectronic emission method) and manufacturing methods for the fabrication of a particular material (or its constitution) are marked
[c]Absolute work functions
[d]The values of effective electron work function
[e]Calculated using the Perdew's generalized gradient approximation (GGA) for the fixed surface model
[f]Calculated using the Perdew's GGA for the relaxed surface model
[g]Full-potential linear-"muffin-tin"-orbital calculation technique (40-atom cell, no vacancies)
[h]For graphene-covered surface φ = 3.9 eV
[i]Calculated using the linear-muffin-tin-orbital with atomic sphere approximation (LMTO-ASA) technique
[j]For graphene-covered surface φ = 3.5 eV

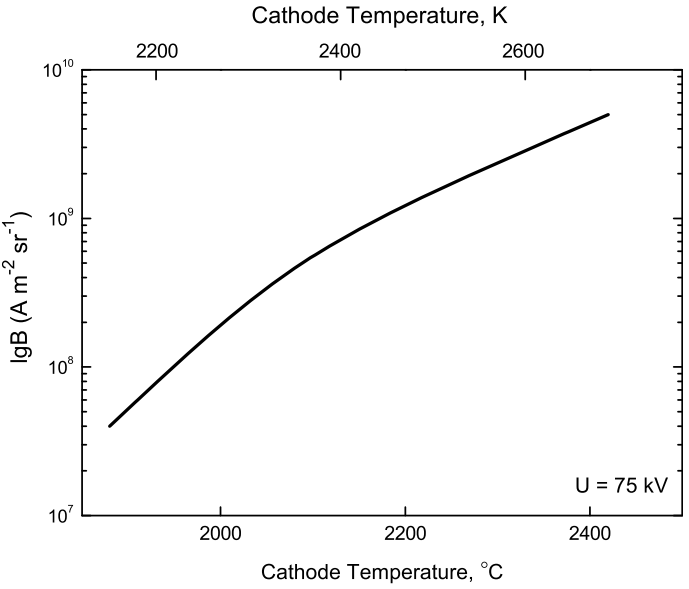

Fig. 2.18 Typical brightness (B) characteristic (U = 75 kV) of tantalum monocarbide cathode (thermionic emitter with electron work functions φ = 4.36 eV, φ_{eff} = 4.33 eV, Richardson constant A = 142.7 × 10^4 A m^{-2} K^{-2}, T = 2635 K) produced by the carburization reaction of metal wire/foil with benzene vapour [150]

The recommended values of electrical resistivity, magnetic susceptibility, integral and spectral emittances and thermionic emission characteristics (electron work function and Richardson constant) for tantalum carbides are given in the wide range of temperatures in comparison with other ultra-high temperature materials in Addendum.

2.4 Physico-mechanical Properties

Most of the physico-mechanical properties of tantalum carbides are sensitive to the deviations from the stoichiometry, but also to the crystallographic directions in the materials; it has been just marked in Sect. 2.1 concerning the slip systems in TaC_{1-x}. At room temperature the hardness *HV*, GPa of near-stoichiometric tantalum monocarbide TaC_{1-x} is evaluated as 11.0 (mean grain size – 4-9 μm, porosity 3–8%, 5–15 N load) [724], 12.4 (for thin films on Ti) [686], 13.7 [403], 13.9 ± 0.7 (mean grain size – 5.8 ± 0.2 μm, porosity < 2%, 9.8 N load) [729], 14.0 [42, 332, 333], 14.0–14.3 (mean grain size – 15-25 μm, porosity < 2%, 98 N load) [721], 14.3 (for $TaC_{0.99}$, 1 N load) [601], 14.6 (for $TaC_{0.96}$, 200 N load) [1, 23], 14.7–19.6 [605], (15.1–17.7) ± (0.7–1.0) [35, 123, 171, 334], 15.2 [345], 15.5 [16], 15.7 (extrapolated to $TaC_{1.0}$) [109, 542], 15.7 ± 0.5 (0.3–49 N load) [16, 61, 90, 892], 16.7 [60], 16.9 ± 1.4 (for $TaC_{0.99}$, 0.5 N load) [37, 604], 17.0 ± 1.0 (contents: O – 0.05%, N – 0.05%, 0.5 N load) [164, 606], 17.7–19.6 (0.5–1.0 N load) [67, 197, 600, 602, 603], 17.8 (0.2–0.5 N load) [67, 600], 18.5 ± 0.8 (for $TaC_{0.98}$) [167] and 19.6 ± 1.0 (for $TaC_{0.95}(N,O)_{0.02}$, porosity – 3 to 5%, 0.5 N load) [168]; hardness *HK*, GPa is 18.0 [165] and 19.1 [387]. The hardness of $TaC_{\sim 1.0}$ in Mohs scale is 9–10 [442, 841], and in Rockwell scale *HRA* it is 82 kgf mm^{-2} (0.80 GPa) [37, 171], ~88 kgf mm^{-2} (~0.86 GPa) [392] and 89 kgf mm^{-2} (0.87 GPa) [388]. The variations of hardnesses *HV*/*HK* and microhardness $H\mu$ with carbon content within the homogeneity range of TaC_{1-x} and the orientation of indenter for the (100) surface of single crystal $TaC_{0.83}$ and $TaC_{0.96}$ are demonstrated in Fig. 2.19. According to Rowcliffe and Warren [381], a phase change occurs at hardness impressions made in TaC_{1-x} compositions in the interval from ~$TaC_{0.75}$ to $TaC_{0.83}$, the maximal hardness in the Ta–C system, ~37–51 GPa (*HK* scale, 1 N load) is corresponding to composition $TaC_{0.83\pm0.01}$. The variation of hardness *HV* of near-stoichiometric tantalum monocarbide TaC_{1-x} with temperature based on several resources is presented in Fig. 2.20. In general, the exponential temperature dependency of hardness for TaC_{1-x} materials in the temperature range from 0 to 540 °C can be given by the following approximate relationship [16, 612–614]:

$$HV = 17.7 \exp[-(6.75 \times 10^{-4})t], \tag{2.52}$$

where *HV* is Vickers hardness, GPa and *t* is temperature, °C. Nino et al. [721] revealed that spark-plasma sintered $TaC_{\sim 1.0}$ materials over the range of grain sizes from 3 to 26 μm did not conform Hall-Petch type relationship. The hardness *HV* of

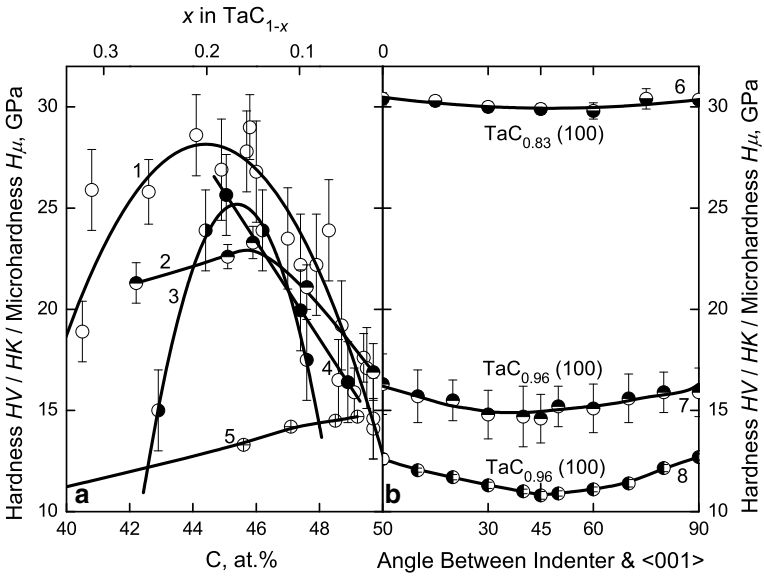

Fig. 2.19 Variations of the hardnesses *HV* (1, 5–6, 8) and *HK* (3–4, 7) and microhardness *Hμ* (2) at room temperature of tantalum monocarbide TaC_{1-x}: (**a**) with deviation from the stoichiometry within the homogeneity range (1 – produced by gas carbidization and hot-pressing methods, 200 N load [23, 332, 333]; 2 – hot-pressed and annealed in vacuum, 6–10% porosity, 0.5 N load [326, 666]; 3 – materials carburized in propane, 1 N load [110, 380, 527]; 4 – carburized by solid-state diffusion saturation, mean grain size – 10 μm, 1 N load [187, 380, 527]; 5 – [109]) and (**b**) with indenter orientation (angle between indenter axis and <001> direction) on the (100) surface of single crystal materials grown by floating-zone processes (6 – $TaC_{0.83}$, 2 N load [169, 170]; 7 – $TaC_{0.96}$, 1 N load [154, 155]; 8 – $TaC_{0.96}$, 2 N load [154, 155])

produced by gas carburization stoichiometric tantalum semicarbide $Ta_{\sim 2.0}C$ at room temperature is ~13.3 GPa [23]; for sintered α-$Ta_{2+x}C$ Kotelnikov et al. [16] and Andrievskii and Spivak [41] recommended the value of 16.8 ± 1.6 GPa (0.3 N load), though Stern [605] and Ellinger [43] reported for hardnesses *HV* and *HK* much lower value of ~9.8 GPa. The data on hardness *HK* of tantalum monocarbide single crystals and microhardness *Hμ* of various tantalum semicarbide and monocarbide compositions are listed in Tables 2.11 and 2.12. The mechanical properties of nanostructured TaC_{1-x} (99.5% purity), including its hardness *HV*, are presented in Table 2.13.

On the basis of several sources the variations of ultimate tensile σ_t and flexural (bending) σ_f strengths of tantalum monocarbide TaC_{1-x} with carbon content, temperature and mean grain size are shown in Figs. 2.21 and 2.22. The relationship $\sigma_f/\sigma_t \approx 2$ can be recommended for general estimations of the strength characteristics of TaC_{1-x} materials [42, 110, 615]. At room temperature the ultimate compressive strength for sintered (mean grain size – 5 to 20 μm, porosity – 5%) $TaC_{0.99}$ is evaluated as 1.75 GPa (at strain rate 3.7×10^{-3} s^{-1}) [1] and for sintered $TaC_{\sim 1.0}$ – 0.84 GPa (corrected for porosity) [663]; the values of transverse rupture strength of

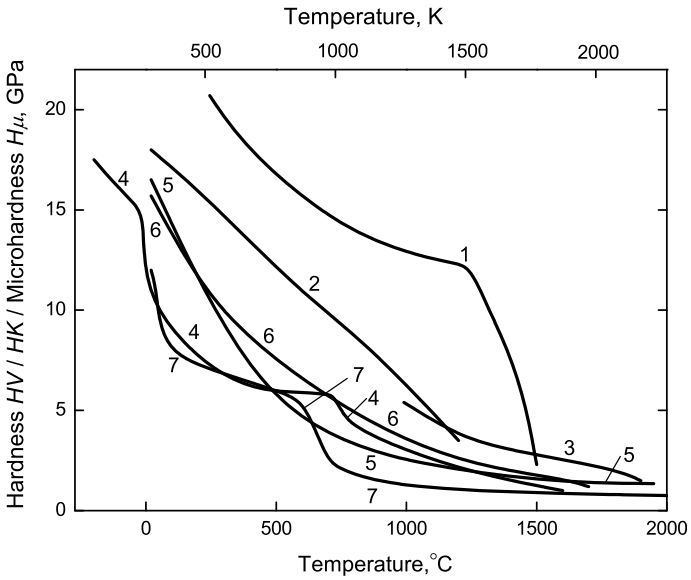

Fig. 2.20 Variation of hardnesses *HV* (2, 4–5, 7), *HK* (3) and microhardness *Hμ* (1, 6) with temperature for TaC_{1-x} materials with various deviations from the stoichiometry based on several resources: 1 – single crystal (prepared by floating-zone process) $TaC_{0.83}$ (100), along the <001> direction, 2 N load [169, 170, 517]; 2 – sintered with temporary metallic binder (in high vacuum) $TaC_{\sim 1.0}$, 3 N load [842]; 3 – arc-cast, 9.8 N load [517, 520]; 4 – sintered in vacuum $TaC_{0.95}$, 1 N load [42]; 5 – hot-pressed [1]; 6 – hot-pressed $TaC_{\sim 1.0}$, porosity – 8%, 0.6–0.8 N load [425]; 7 – sintered [216] (when it is not indicated specially, data given are for quasi-stoichiometric composition)

spark-plasma sintered $TaC_{\sim 1.0}$ (mean grain size – 4-9 μm, porosity 3–8%), measured by applying pressure through a ball-shaped indentor onto the disc-shaped samples, vary from 336 to 372 MPa [724]. The near-stoichiometric monocarbide TaC_{1-x} materials exhibit average fracture toughness K_{IC} of 5.6 ± 1.8 [704, 762], 3.8± 0.4 [721], 3.1 [771] and 2.7 ± 0.3 MPa $m^{1/2}$ [729] – for spark-plasma sintered materials, 5.3 ± 0.7 [34], 4.1 [892] and 3.7 ± 0.2 MPa $m^{1/2}$ [728] – for hot-pressed materials and 3.2 MPa $m^{1/2}$ – for hot isostatically pressed materials [398]. The fracture toughness was found to decrease with decreasing carbon content within the homogeneity range, as for $TaC_{0.8-0.9}$ compositions it was 3.8 ± 0.1 (for hot-pressed materials) [34] and 3.4 ± 0.9 MPa $m^{1/2}$ (for vacuum plasma sprayed materials) [181]; however, the opposite trend was revealed for materials prepared by hot-isostatic pressing, as for the same compositions K_{IC} increased up to 4.7 MPa $m^{1/2}$ [398]. For 99.5% purity nanostructured TaC_{1-x} with mean grain size in the range of 30–180 nm K_{IC} increases with decreasing the values of porosity and mean grain size (*see* Table 2.13). The exceptionally high values of fracture toughness were reported for two-phase carbide compositions: hot-pressed $TaC_{\sim 0.6}$ materials containing ~80% ζ-Ta_4C_{3-x}, K_{IC} = 12.7 ± 0.7 MPa $m^{1/2}$ [34] and $TaC_{\sim 0.7}$ materials,

Table 2.11 Microhardness $H\mu$, of tantalum monocarbide TaC_{1-x} ($0 < x \leq 0.17$) single crystals at room temperature (1 N load)

Surface	Indentor diagonal direction	Microhardness $H\mu$, GPa	References
		$x \approx 0$	
(001)	<001>	14.2 ± 0.5	[1, 159]
	<100>	16.2	[838, 839]
	<110>	14.7	[838, 839]
	<110>	12.8 ± 0.5	[1, 159]
	–	24.0 ± 0.5	[1, 160]
	–	22.1 ± 0.5[a]	[1, 160]
(110)	–	20.2 ± 0.5	[1, 160]
	–	19.1 ± 0.5[a]	[1, 160]
(111)	$<1\bar{1}0>$	13.1 ± 0.5	[1, 159]
	$<11\bar{2}>$	14.9 ± 0.5	[1, 159]
	–	19.1 ± 0.5	[1, 160]
	–	18.4 ± 0.5[a]	[1, 160]
		$x = 0.01 - 0.04$	
(100)	<010>	14.9 ± 0.5	[381]
		$x = 0.04$	
(001)	<100>	16.3 ± 0.9[b]	[154, 155]
		12.6 ± 0.3[c]	[154, 155]
	<010>	16.1 ± 0.5[b]	[154, 155]
		10.8 ± 0.2[c]	[154, 155]
	<110>	14.4 ± 0.5[b]	[154, 155]
		12.7 ± 0.2[c]	[154, 155]
(100)	<010>	15.2 ± 1.0[b]	[381]
		$x = 0.17$	
(100)	<001>	30.4 ± 0.3[c]	[169, 170]
	<010>	~38[b]	[381]
	<010>	30.4 ± 0.3[c]	[169, 170]
	<011>	29.9 ± 0.3[c]	[169, 170]

[a]Annealed at 2000 °C
[b]Knoop measurement (*HK*)
[c]2 N load

also containing lamellar ζ-phase, K_{IC} = (11.9–13.8) ± (0.8–0.9) MPa $m^{1/2}$ [523]. The stoichiometric semicarbide Ta_2C formed by hot-isostatic pressing was found to have K_{IC} of 9.7 ± 0.8 MPa $m^{1/2}$ [406]; for the spark plasma sintered semicarbide materials K_{IC} = (4.8–6.4) ± (0.6–0.7) MPa $m^{1/2}$ [405]. The mechanical properties, including hardness *HV*/*HK*, flexural strength σ_f and fracture toughness K_{IC}, of tantalum semicarbide $Ta_{2\pm x}C$ materials prepared by different methods are given in Table 2.14. The ductile-to-brittle transition temperature of near-stoichiometric TaC_{1-x} is approximately half its melting point [178]; the reported values for the transition temperature of phases with various deviations from the stoichiometry are listed in Table 2.15. Tantalum semicarbide α-$Ta_{2+x}C$ is brittle at room temperature but much more ductile than TaC_{1-x} in the range of temperatures around 1900 °C [189, 517, 524]. For detailed data on the slip systems in tantalum carbide phases

Table 2.12 Microhardness $H\mu$ of the various compositions of tantalum carbide phases at room temperature

Composition	Microhardness $H\mu$, GPa	Remarks	References
Ta_2C	~18	0.5 N load, physically vapour deposited (Ta in the presence of C_2H_2) coatings	[400]
	15.5 ± 0.7[a]	Hot isostatically pressed	[406]
	16.8 ± 1.6	0.3 N load, sintered	[37, 215]
	12.0	Sintered	[1, 161]
	9.4	–	[109, 517]
$TaC_{0.73}$	21.7 ± 1.0	0.5 N load, sintered in vacuum, contents: O – 0.12%, N – 0.06%	[1, 37]
$TaC_{0.80}$	~30	–	[42, 332]
	20.0 ± 0.5	9.8 N load, hot-pressed, mean grain size – 5.8 ± 2.4 μm, porosity – 3.3%	[34]
$TaC_{0.8–0.9}$	25.0 ± 7.2[b]	0.5 N load, vacuum plasma sprayed, mean grain size – 60-90 nm	[181]
$TaC_{0.82}$	23.1 ± 0.6	0.5 N load, sintered in vacuum, contents: O – 0.12%, N – 0.02%	[1, 37]
$TaC_{0.83}$	29.4	0.25 N load, materials carburized in propane (maximal hardness revealed within the homogeneity range)	[110]
	23.5	1 N load, materials carburized in propane (maximal hardness revealed within the homogeneity range)	[110]
$TaC_{0.84}$	13.3	–	[109, 517]
$TaC_{0.85}$	23.8 ± 0.8	0.5 N load, sintered in vacuum, contents: O – 0.09%, N – 0.04%	[37]
$TaC_{0.89}$	14.2	–	[109, 517]
$TaC_{0.90}$	18.0 ± 0.4	9.8 N load, hot-pressed, mean grain size – 6.4 ± 3.7 μm, porosity – 3.5%	[34]
$TaC_{0.91}$	22.4[a]	0.5 N load, physically vapour deposited (Ta in the presence of C_2H_2) coatings	[400]
	21.5 ± 1.1	0.5 N load, sintered in vacuum, contents: O – 0.14%, N – 0.06%	[1, 37]
$TaC_{0.93}$	24.3[a]	0.5 N load, physically vapour deposited (Ta in the presence of C_2H_2) coatings	[400]
	15.1 ± 0.7	0.2-1.5 N load, hot-pressed and annealed at 1800 °C for 3 h, porosity – 20%, content: non-combined C – 0.3%	[171, 379]
$TaC_{0.94}$	15.7 ± 0.5	–	[90]
$TaC_{0.95}$	19.6 ± 1.0	0.5 N load, hot-pressed, porosity – 3-5%, contents: O + N – 0.02%	[168]
	10.5	1.0 N load, sintered in vacuum	[1, 42]

(continued)

Table 2.12 (continued)

Composition	Microhardness $H\mu$, GPa	Remarks	References
$TaC_{0.96}$	16.8	1.0 N load, sintered in vacuum, annealed, mean grain size – 15 μm, porosity – 4.5%, contents: non-combined C – 0.05%, O – 0.46%, N – 0.02%	[1, 41]
$TaC_{0.97}$	17.6[a]	0.5 N load, physically vapour deposited (Ta in the presence of C_2H_2) coatings	[400]
	14.7	–	[109, 517]
$TaC_{0.98}$	19.25	Deposited from gas phase	[163]
	18.5 ± 0.8	–	[1, 167]
	18.0[a]	Hot-pressed, mean grain size – 0.2 μm	[162]
	15.2 ± 0.7	Pyrolytic materials	[610]
	13.0[a]	Hot-pressed, mean grain size – 4.0 μm	[162]
$TaC_{0.99}$	17.2 ± 1.4	1.0 N load, sintered in vacuum, contents: non-combined C – 0.12%, O – 0.11%, N – 0.03%	[1, 37]
	14.3[c]	1.0 N load, sintered in vacuum, annealed, mean grain size – 12 μm, porosity – 4%, contents: non-combined C – 0.08%, O – 0.05%, N – 0.005%	[1, 41]
$TaC_{\sim 1.0}$	25.0	–	[115, 544, 550, 551]
	24.5	DFT-computed theoretical value	[435]
	24.2[a]	The highest value for materials sintered in vacuum (with fugitive binder), porosity – 7.5%	[387]
	21.3	DFT-computed theoretical value	[618]
	21.4	DFT-computed theoretical value	[844]
	19.9	Calculated on the basis of Simunek-Vackar's model	[383–385]
	19.5	–	[67]
	19.1[a,d]	0.5 N load	[61, 387]
	18.9 ± 3.4	20-30 N load, spark-plasma sintered, porosity – 5.4 ± 0.7%, mean grain size – 4.5 ± 0.9 μm	[704, 762]
	18.8[a]	0.05 N load, hot-pressed, porosity – 5%	[335]
	18.0[a]	Hot-pressed from ultra-dispersed powder	[1, 165]
	18.0	Hard coatings	[556, 557]
	18.0	3 N load, sintered with temporary metallic binder in high vacuum	[842]
	17.9	Sintered	[545, 548]
	17.6–19.6	0.2-0.5 N load	[67, 68]
	17.6–17.8	0.2 N load	[166, 174]
	17.6	0.5 N load	[600]
	17.0 ± 1.0	0.5 N load, contents: O – 0.05%, N – 0.05%	[1, 164]
	17.0	Calculated on the basis of modified Hall-Petch relationship theory for nanocrystals	[324]
	16.7[e]	Deduced from the dislocation data measured by X-ray diffraction profile analysis	[558]
	16.7[f]	0.2–10 N load, hot-pressed, porosity < 5%	[60, 535]

(continued)

Table 2.12 (continued)

Composition	Microhardness $H\mu$, GPa	Remarks	References
	16.1[g]	0.04–0.12 N load (extrapolated to 0.01 N load), coating ion-sputtered on a glass substrate, thickness – ~0.2 μm	[324, 558]
	16.0	1.2 N load	[61]
	15.7[h]	On the basis of several published sources (extrapolated to defectless theoretical stoichiometry)	[109, 324, 558]
	~15.7[i]	0.3 N load	[16, 61]
	15.5	0.5 N load	[67]
	15.5 ± 0.6	2 N load, spark-plasma sintered, porosity – ~7%	[771]
	15.2	–	[840]
	(15.1–17.6) ± (1.0–1.7)	–	[35, 123, 171, 334]
	~15.0	1 N load, single-crystal materials	[381, 663]
	14.7	Calculated on the basis of Xue's model	[383, 386]
	14.6[j]	0.5–3.9 N load, hot filament chemical vapour deposited films on metal Ta substrate, thickness – 8 μm	[401]
	14.1 ± 0.2	5 N load, hot-pressed, porosity – 6%, mean grain size – 3 μm	[402]
	(14–19) ± (1–3)	Pulsed laser ablation deposited thin films on Ti substrate, thickness – 0.24-0.25 μm, approximate grain size – 10-100 nm	[611]
	~14.0	–	[42, 332]
	13.7–21.2[k]	0–0.3 N load, chemical vapour deposited coatings, thickness 60–80 μm	[409]
	13.7[l]	–	[403]
	13.5 ± 0.2	9.8 N load, hot-pressed, mean grain size – 7.7 ± 4.5 μm, porosity – 3%	[34]

[a]Knoop measurement (*HK*)
[b]Average for three directions
[c]3.8 GPa – at 1000 °C
[d]Probably contaminated by W
[e]For ion-sputtered coating the measured dislocation density $\rho = 1.65 \times 10^{12}$ cm^{-2}
[f]Determined in accordance with the formula: $H_\mu = 18.18P/d^2$, where P is the load on the indentor, N and d is the diagonal of the indentation, μm; the values of other micromechanical characteristics: microbrittleness $\gamma_\mu = (D^2 - d^2)/d^2 = 0.55$, where D is the average size of the damageability zone, μm and brittle microstrength $\sigma_\mu = P/D^2 = 5.7$ GPa [535]
[g]Measured using a Vickers diamond microindentation hardness tester
[h]Recommended for bulk samples
[i]8.0 GPa – at 730 °C, 4.5 – at 1230 °C [16, 166]
[j]Measured using a Wilson microhardness tester
[k]Nanoindentation with a typical probe depth of 1 μm
[l]Diamond pyramid hardness measurement (*DPH*)

Table 2.13 Mechanical properties of nanostructured tantalum monocarbide TaC_{1-x} (99.5% purity)[a] at room temperature [172, 203]

Relative density, %	Mean grain size, nm	Hardness HV[b], GPa	Fracture toughness K_{IC}^{c}, MPa $m^{1/2}$
88	184	13	3.9 ± 0.3
95	72	14	4.1 ± 0.4
96	33	21–22	5.1–6.8

[a]Manufactured by high-frequency induction heated sintering method
[b]196 N load
[c]Estimated according to $K_{IC} = 0.016P/C^{3/2}(E/HV)^{1/2}$, where P is the indentation load, C is the trace length of the crack measured from the centre of indentation and E is Young's modulus

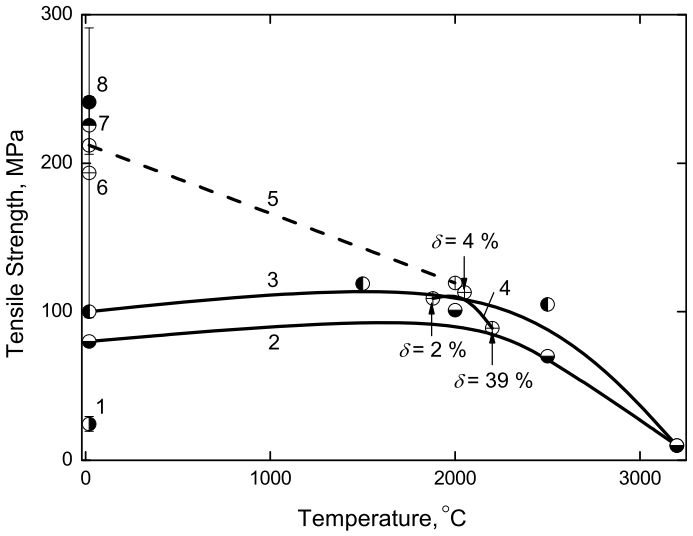

Fig. 2.21 Variations of tensile strength of near-stoichiometric tantalum monocarbide TaC_{1-x} with temperature on the basis of several sources: 1 – [35, 61]; 2 – sintered, 30% porosity [1, 173]; 3 – sintered, 20% porosity [1, 173]; 4 – carburized by solid-state diffusion saturation, mean grain size – 10 μm [123, 187, 389]; 5 – [61, 174]; 6 – hot-pressed $TaC_{0.93}$, 2–6% porosity [392, 396]; 7 – [61, 102]; 8 – hot-pressed $TaC_{0.95}$, mean grain size – 13 μm, porosity – 7% [396, 540, 560] (elongation at rupture δ is indicated for *curve* 4)

structures *see* Sect. 2.1. The values of prolonged strength (creep resistance) are of great importance for the evaluation of operational time of components in the engineering practice. The examples of intermediate-temperature creep characteristics of hot isostatically pressed $TaC_{0.99}$ (porosity – 3%, mean grain size – 57 μm) [182] are demonstrated in Fig. 2.23, high- and ultra-high temperature creep characteristics of various compositions TaC_{1-x} – in Fig. 2.24; steady-state creep activation energies Q and exponent constants n of TaC_{1-x} phases are listed in Table 2.16. The high-temperature creep mechanisms in tantalum semicarbide differ considerably from those in the monocarbide phase. Semicarbide α-$Ta_{2+x}C$ phase

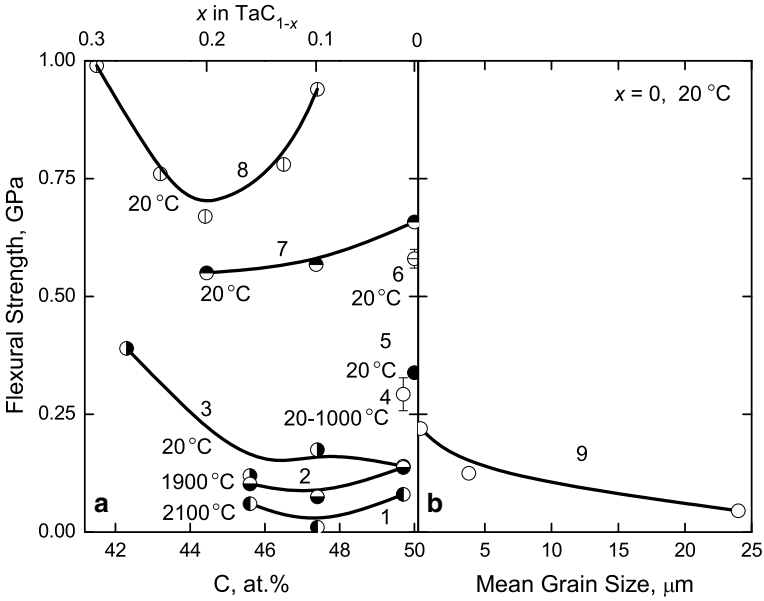

Fig. 2.22 Variations of flexural (bending) strength of TaC_{1-x} materials on the basis of several sources: (**a**) with deviation from the stoichiometry and temperature (1-3 – sintered [175, 737]; 4 – hot-pressed, 7–20% porosity, non-combined C – 0.03-0.07%, 3-point bending scheme [394, 395]; 5 – sintered at 2300 °C, mean grain size ~2 μm, 2.5% porosity [176]; 6 – hot-pressed, mean grain size – 4.0 ± 0.8 μm, 6% porosity, 4-point bending scheme [728]; 7 – hot-pressed, mean grain size – 7 ± 3 μm, 3% porosity [34]; 8 – materials carburized in propane, 3-point bending scheme [110, 380, 527]) and (**b**) with mean grain size (9 – sintered in H_2 at 2700 °C [177])

exhibits significantly more plasticity compared to TaC_{1-x}. The graphic stress-strain diagrams (curves) for different tantalum carbide materials at 1930 and 2160 °C are given in Fig. 2.25. Approximately, the following main dominating creep mechanisms can be established in some overlapping parametric ranges for tantalum monocarbide TaC_{1-x} materials with certain changes within the homogeneity range [183]:

vacancy-diffusional (or Nabarro-Hering) mechanism controlled by the bulk diffusion of metal in carbide (at temperatures from 2400 to 3000 °C and applied stresses from 8 to 70 MPa, with $Q \approx 600$–640 kJ mol^{-1} and $n = 1$);
dislocation-diffusional mechanism controlled by the bulk diffusion of metal in carbide (at 2500–3000 °C and 20–70 MPa, with $Q \approx 600$–670 kJ mol^{-1} and $n \approx 3.0$–4.5);
viscous dislocation sliding mechanism controlled by the bulk diffusion of carbon in carbide (at 1600–2200 °C and 50–200 MPa, with $Q \approx 400$–500 kJ mol^{-1} and $n \approx 3.5$–5.0).

At room temperature the values of Young's E, Coulomb's (shear) G and bulk (compression) K moduli, volume compressibility κ and Poisson's ratio v of sub- and

Table 2.14 Mechanical properties of tantalum semicarbide $Ta_{2\pm x}C$ materials prepared by different manufacturing methods at room temperature [398, 405, 406]

Relative density, %	Mean grain size, μm	Hardness *HV*/*HK*, GPa	Flexural strength σ_f, MPa	Fracture toughness K_{IC}, MPa $m^{1/2}$	Young's modulus *E*, GPa
95.8[a,b]	1.3	11.7 ± 0.8[c]	461 ± 55[d]	4.8 ± 0.7[e]	489 ± 11[f]
95.9[a,g]	3.7	11.6 ± 0.8[c]	548 ± 52[d]	6.0 ± 0.6[e]	484 ± 12[f]
97.5[a,g]	18	11.8 ± 0.9[c]	507 ± 45[d]	6.4 ± 0.7[e]	493 ± 11[f]
>98[h]	–	15.5 ± 0.7[i]	550 ± 50	(9.3–9.7) ± 0.8[j]	–

[a]Manufactured by spark plasma sintering method
[b]Open porosity – 0.3% (density – 14.42 g cm^{-3})
[c]Vickers hardness *HV*, 49 N load
[d]3-point bending scheme (fracture direction – normal to the pressing direction)
[e]Estimated according to the equation $K_{IC} = AP/C^{3/2}(E/HV)^{1/2}$, where *A* is a material-independent constant, *P* is the indentation load, *C* is the trace length of the crack measured from the centre of indentation and *E* is Young's modulus
[f]Caculated by the stress-strain curves recorded from tensile surfaces of the bending samples
[g]No open porosity (density – 14.43-14.66 g cm^{-3})
[h]Manufactured by HIP-reaction sintering in a metal Ti container
[i]Knoop microhardness
[j]Measured by conventional single edge notched beam (SENB) method

Table 2.15 Ductile-to-brittle transition temperatures of tantalum monocarbide phases TaC_{1-x}[a,b]

Composition	Temperature, K (°C)	Characteristics	References
$TaC_{\sim 1.0}$	2000 (1730)[c]	Hot-pressed, annealed (porosity – ~10%)	[42, 178]
	2020 (1750)	Carburized by solid-state diffusion saturation	[187, 389]
	2125 ± 100 (1850 ± 100)	On the basis of several sources	[517, 519]
	2300 (2030)	Carburized by solid-state diffusion saturation	[42]
$TaC_{0.98}$	2100–2200 (1830–1930)	Hot-pressed, annealed (porosity – 5-10%)	[175, 179]
	1170 (800)	Carburized, polycrystalline (bending scheme)	[437–439]
$TaC_{0.95}$	1550 (1280)[d]	Hot-pressed (mean grain size – 13 μm, porosity – 7%)	[540]
$TaC_{0.90}$	2025–2125 (1750–1850)	Hot-pressed, annealed (porosity – 5-10%)	[175, 179]
$TaC_{0.78}$	~1600 (1330)	–	[179, 180]
	1870 (1600)	–	[518]

[a]For near-stoichiometric TaC_{1-x} the following average values were estimated: Peierls stress τ_P = 4.0 GPa, shear modulus – Peierls stress ratio τ_P/G = 0.021 and slip activation energy barrier q = 1.7 eV [562]
[b]Transition temperature decreases almost linearly with the value of x growing [107, 175, 380]
[c]In 4-point bending scheme only ~0.1% strain could be achieved prior to failure at 2000 °C
[d]At strain rates ≤8×10^{-4} s^{-1}; permanent strains >5% achieved at 1280 °C and up to 40% at ≥1640 °C

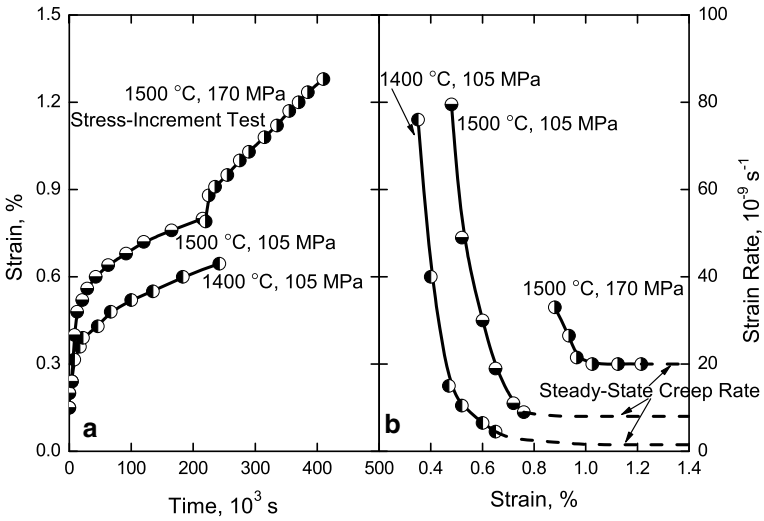

Fig. 2.23 Intermediate-temperature creep characteristics of hot isostatically pressed tantalum monocarbide $TaC_{0.99}$ (porosity – 3%, mean grain size – 57 μm) at 1400–1500 °C and 105–170 MPa: (**a**) strain as a function of time and (**b**) strain rate as a function of strain (extrapolation to assumed values of steady-state creep rates is shown) [182]

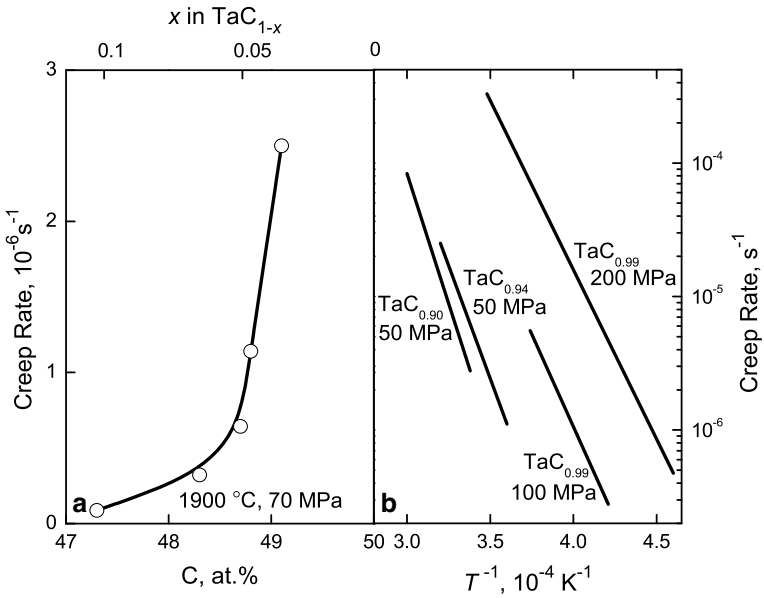

Fig. 2.24 High- and ultra-high temperature steady-state creep rate variations of tantalum monocarbide TaC_{1-x} materials: (**a**) – with deviation from the stoichiometry (materials carburized by solid-state diffusion saturation, mean grain size – 10 μm [187]) and (**b**) – with temperature ($TaC_{0.90}$ sintered and annealed in high-purity He atmosphere, porosity – 7%, mean grain size – 30 μm [186]; $TaC_{0.94}$ sintered in vacuum, porosity – ≤3%, mean grain size – 27 μm, contents O + N – 0.13% [10]; $TaC_{0.99}$ sintered, porosity – 3%, mean grain size – 5 μm, contents: non-combined C – 0.20%, O – 0.18%, N – 0.01%, W – 0.10% [422])

Table 2.16 Formal creep characteristics (activation energy Q, stress exponent constant n) of tantalum monocarbide phases TaC_{1-x} at various temperature and applied stress ranges

Composition	Load type[a]	Temperature range, °C	Applied stress range, MPa	Activation energy[b], Q, kJ mol^{-1}	Exponent constant[b], n	References
$TaC^{c}_{0.99–1.0}$	C	2350–2700	8–70	~610–630	1	[183]
$TaC^{d}_{0.99}$	C	1400–1500	105–170	420	2	[182]
$TaC^{e}_{0.99}$	C	2000–2400	60–100	485	3.5	[422–424]
	C	2000–2400	100–200	485	5	[422–424]
	C	2400–2600	100–200	615–630	4.2	[422–424]
$TaC_{0.98}$	F	1200–2200	–	560–600	–	[437, 438]
$TaC^{f}_{0.97}$	F	2500–3000	0–15	610	1	[184]
$TaC^{g}_{0.97}$	C	2700–3000	20–30	665	3	[534]
$TaC^{h}_{0.95}$	F	2500–2850	7–12	610 ± 30	1	[168]
$TaC^{i}_{0.95}$	C	1280–1640	–	~375	~13	[539, 540]
	C	1460–1810	–	400	4.5	[540, 683]
$TaC^{j}_{0.94}$	C	2500–3050	50	800	–	[10]
$TaC^{k}_{0.90}$	C	2550–3100	50	770	–	[185, 186]
$TaC^{l}_{0.85–0.98}$	T	1950–2100	45–70	710	2–4	[187, 426]
$TaC^{m}_{0.83}$	H	1200–1500	–	465	4.3	[169, 170]

[a]Denoted: T – tension, F – flexure (bending), C – compression, H – hot hardness (measurement)

[b]For Q and n *see* the Eq. (I-3.17) and consideration in [683]

[c]Sintered materials, for $TaC_{0.99}$ (porosity – 2-5%, mean grain size – 4-16 μm) Q = 632 kJ mol^{-2}, for $TaC_{\sim 1.0}$ (mean grain size – 10 μm) Q = 607 kJ mol^{-2} and for $TaC_{\sim 1.0}$ (mean grain size – 16 μm) Q = 615 kJ mol^{-2}

[d]Hot isostatically pressed (HIP), porosity – 3%, mean grain size – 57 μm

[e]Sintered, porosity – 3%, mean grain size – 5 μm, contents: non-combined C – 0.20%, O – 0.18%, N – 0.01%, W – 0.10%

[f]Hot-pessed, porosity – 3-5%

[g]Characteristics were determined for κ-creep (during hot-pressing densification of submicrometer powders) using Kovalchenko's equation for inhomogeneous deformation of viscous porous body: $\Delta X_n(\rho) = \int_{\rho_0}^{\rho} \frac{\rho^{\frac{5n+1}{2}}}{(1-\rho)^{\frac{n+1}{2}}} d\rho = B(P + P_L)^n \tau$, where $\Delta X_n(\rho)$ is the difference between the primitives of the functions $X_n(\rho)$ taking for the current ρ and initial ρ_0 values of relative density, P is the applied pressure, P_L is the Laplace pressure, B and n are corresponding to those in Eq. (I-3.17)

[h]Hot-pressed, porosity – 3-5%, mean grain size – 33 μm, contents: (O + N)/Ta ratio is 0.02

[i]Hot-pressed in vacuum and annealed, porosity – 7%, mean grain size – 13 μm, characteristics were determined for the mechanical equation of state at the yield stress $d\gamma/dt = A(\tau_{0.2}/G)^n \exp(-Q/RT)$, where $d\gamma/dt$ is the plastic shear strain rate, which equals the true plastic strain rate divided by the Schmid factor; $\tau_{0.2}$ is the critical resolved shear stress, corresponding to 0.2% proof stress; G is Coulomb's (shear) modulus; R is the gas constant and T is absolute temperature, K

[j]Sintered in vacuum, porosity – $\leq$3%, mean grain size – 27 μm, contents: O + N – 0.13%

[k]Sintered, porosity – 7%, mean grain size – 30 μm

[l]Carburized by solid-state diffusion saturation, mean grain size – 10 μm

[m]Single crystal prepared by floating-zone process, plane direction – (100) <001>

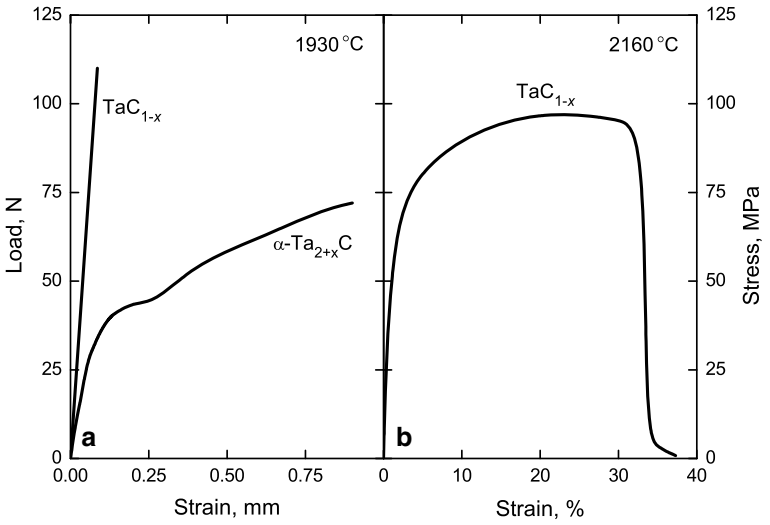

Fig. 2.25 The examples of stress-strain diagrams (curves) of tantalum carbides materials at high temperatures: (**a**) mid-point deflection at 4-point flexural (bending) scheme, 1930 °C, semicarbide α-$Ta_{2+x}C$ and monocarbide TaC_{1-x}, prepared under similar conditions by hot isostatic pressing (HIP) [189] and (**b**) tensile scheme (elongation to failure – 33%), 2160 °C, monocarbide TaC_{1-x}, prepared by carburization (solid-state diffusion saturation), mean grain size – 10 μm [389, 517]

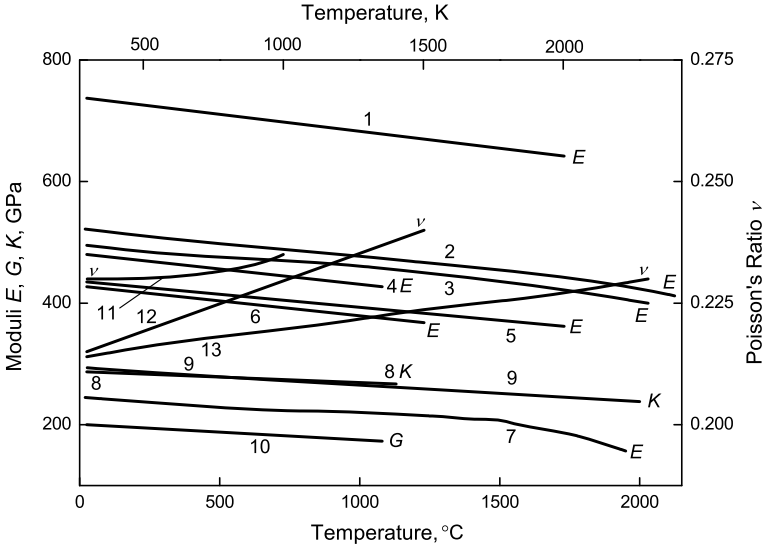

Fig. 2.26 Variations of elastic properties (1-7 – Young's modulus E, 8-9 – bulk (compression) modulus K, 10 – Coulomb's (shear) modulus G and 11-13 – Poisson's ratio ν) of near-stoichiometric tantalum carbide phases with temperature: 1 – monocarbide TaC_{1-x} and 5 – semicarbide α-$Ta_{2+x}C$ (recommended by Kotelnikov et al. [16]); 2 – $TaC_{0.99}$ (measured by Bukatov et al. [50]); 3 and 13 – $TaC_{0.95}$ with 3% porosity and contents N – 0.27% (given by Andrievskii and Spivak [41]); 4, 6, 8, 10, 12 – $TaC_{0.99}$ with 1% (4, 8, 10), 7% (12) and 12% (6) porosity (measured and calculated by Jun and Shaffer [190]); 7 – $TaC_{\sim 1.0}$ (calculated on the basis of hardness-elasticity relationship by Travushkin et al. [216]); 9 – $TaC_{\sim 1.0}$ (theoretically calculated by Cao et al. [413]); 11 – $TaC_{\sim 1.0}$ (theoretically calculated by Varshney et al. [414, 415])

near-stoichiometric tantalum monocarbide TaC_{1-x} materials lie within the areas: E = 240–580 GPa [1, 35, 37, 41, 42, 50, 60, 61, 68, 90, 115, 123, 427, 544], G = 200–250 GPa [41, 42, 60, 192, 397, 407, 427], K = 220–400 GPa [41, 42, 192, 397, 407, 408], κ = 2.84–3.33 TPa^{-1} [41, 42, 195, 336] and ν = 0.20–0.25 [41, 42, 192, 338, 427], while for near-stoichiometric tantalum semicarbide $Ta_{2\pm x}C$ materials E = 440–530 GPa [16, 404, 405] (*see also* Table 2.14), and calculated on the basis of density-functional theory (DFT) for α-$Ta_{2+x}C$: G = 148 GPa, K = 267 GPa [435], K = 246 GPa [428] and for β-$Ta_{2\pm x}C$: K = 229 GPa [428]; the experimentally measured velocities of ultrasonic waves propagated in tantalum monocarbide TaC_{1-x} phases are

for single crystal $TaC_{0.90}$ [191]:

longitudinal velocity (along <100>) V_S, m s^{-1}	5880 ± 70,
transversal velocity (along <110> and <100>) V_T, m s^{-1}	2320 ± 100,
longitudinal velocity (along <110>) V_S, m s^{-1}	5020 ± 70,
transversal velocity (along <110>) V_T, m s^{-1}	3760;

for hot-pressed $TaC_{0.98\pm0.12}$ (porosity – 11.7 ± 1.4%) [206]:

longitudinal velocity V_S, m s^{-1}	6232 ± 10,
transversal velocity V_T, m s^{-1}	3744 ± 13;

for $TaC_{\sim1.0}$ [429] (summarized on the basis of several sources):

average velocity V_m, m s^{-1}	4720;

and theoretically estimated sound velocities for stoichiometric tantalum monocarbide $TaC_{\sim1.0}$ are

by *ab initio* calculations (for zero pressure):

longitudinal velocity V_S, m s^{-1} 6518 (0 K) [414], 5172 (~300 K) [414], 5200 [419], 6190 [618], 6607 [716];
transversal velocity V_T, m s^{-1} 3850 (0 K) [414], 3057 (~300 K) [414], 3170 [419], 3517 [618], 3954 [716];
average velocity V_m, m s^{-1} 3490 [419], 3910 [618], 4369 [716];

by LMTO (linearized method of "muffin-tin" orbitals) calculations [429]:

average velocity V_m, m s^{-1}	4480;

by TBFP (three body force potential) modeling calculations [419]:

longitudinal velocity V_S, m s^{-1}	6240;
transversal velocity V_T, m s^{-1}	3740;
average velocity V_m, m s^{-1}	4130.

The variations of the elastic properties of near-stoichiometric tantalum carbide phases with temperature are shown in Fig. 2.26, for TaC_{1-x} the average rate of decay of the elastic modulus with temperature $1/E_{298K} \times dE/dT = -1.5 \times 10^{-4}$ K^{-1} [562]; the elastic properties (stiffness coefficients c_{ij} and moduli E, K and Poisson's ratio ν) of various tantalum monocarbide TaC_{1-x} phases are listed in Table 2.17. The DFT-calculated stiffness coefficients c_{ij} for α-$Ta_{2+x}C$ are c_{11} = 479, c_{33} = 500,

Table 2.17 Elastic properties (stiffness coefficients c_{11}, c_{12} and c_{44}, Young's modulus E, bulk (compression) modulus K, Poisson's ratio v)[a] of tantalum monocarbide TaC_{1-x} phases

Composition	Stiffness coefficients c_{ij}			Young's modulus E, GPa	Bulk modulus K, GPa	Poisson's ratio v	References
	c_{11}, GPa	c_{12}, GPa	c_{44}, GPa				
$TaC^{b}_{0.75}$	579	172	116	380[c]	308[c]	–	[434]
$TaC^{b}_{0.875}$	684	147	157	488[d]	337[d]	–	[434]
$TaC_{0.90}$	505[e]	73[e]	79[e]	477[f]	229[f]	0.15[f]	[115, 191]
	500[g]	90[g]	80[g]	–	–	–	[208]
	–	–	–	303	–	–	[115, 212]
	320	84	65	–	–	–	[41]
$TaC^{h}_{0.95}$	–	–	–	270	–	–	[396, 540]
$TaC^{i}_{0.96}$	–	–	–	527 ± 10	–	–	[1]
$TaC_{0.97}$	610	210	230	529[f]	370[f]	0.26[f]	[155]
	709[b]	147[b]	169[b]	515[b,j]	337[b,j]	–	[434]
$TaC^{k}_{0.98}$	–	–	–	567 ± 68[l,m]	332 ± 29[l,m]	0.215 ± 0.02[l]	[206]
	–	–	–	–	345 ± 9[n]	–	[198]
$TaC_{0.99}$	–	–	–	550 ± 10[l,o]	–	–	[1, 554]
	–	–	–	552[i]	317[i]	0.21[i]	[50, 432]
	631[p,q]	168[p,q]	–	560[p,q,r]	329[p,q,r]	0.21[p,q]	[190]
	634[s]	200[s]	–	538[l,t]	345[l,t]	0.24[l]	[192, 427]
	–	–	–	–	343	–	[536]
$TaC_{\sim 1.0}$	347[s]	110[s]	–	294	188[f]	0.24	[207]
	595[u]	–	153[u]	–	–	–	[193, 562]
	621[v]	155[v]	167[v]	550[v]	318[v]	0.21[v]	[193]
	–	–	–	–	365[w]	–	[115, 193]
	880[w]	156[w]	410[w]	–	398[w]	–	[199]
	733[v]	112[v]	315[v]	–	319[v]	–	[199]
	–	–	–	–	342[x]	–	[210]
	–	–	–	–	433 ± 7[y]	–	[200]
	907[w]	108[w]	211[w]	–	371[w]	–	[200]
	723[v]	106[v]	177[v]	–	310[v]	–	[200]
	611[w]	243[w]	159[w]	439[w,z]	365[w,z]	0.30[w]	[201]
	466[a1]	224[a1]	121[a1]	320[a1]	305[a1,a2]	0.33[a1]	[201]
	463[a3]	223[a3]	121[a3]	320[a2,a3]	303[a2,a3]	0.32[a3]	[201]
	454[v]	227[v]	120[v]	312[v,a4]	302[v,a4]	0.33[v]	[201]
	550[a5]	150[a5]	190[a5]	479	283	–	[208, 562]
	–	–	–	514[a6,a7]	317[a6,a7]	0.23[a6]	[209]
	740[x]	165[x]	176[x]	–	357[x,a8]	–	[211]
	–	–	–	480	–	–	[212]
	641[w]	146[w]	156[w]	470[w,a9]	311[w,a9]	0.24[w]	[204]
	–	–	–	288[b1]	–	–	[196]
	–	–	–	534[b1]	–	–	[197]
	–	–	–	562[b1]	–	–	[194]
	–	–	–	–	309[b2]	–	[195]
	–	–	–	265–275[b3]	–	–	[202]
	–	–	–	–	324[b4]	–	[339, 430]
	–	–	–	–	330[b5]	–	[359]
	–	–	–	285[b1]	–	–	[61, 214, 215, 436, 843]
	–	–	–	314[b1]	–	0.17[b1]	[388]

(continued)

Table 2.17 (continued)

Composition	Stiffness coefficients c_{ij}			Young's modulus E, GPa	Bulk modulus K, GPa	Poisson's ratio ν	References
	c_{11}, GPa	c_{12}, GPa	c_{44}, GPa				
	562[b6]	159[b6]	146[b6]	492[b6,b7]	294[b6,b7]	0.22[b6]	[416]
	571[b8,b9]	169[b8,b9]	195[b8,b9]	487[b8,b9,c1]	303[b8,b9,c1]	0.23[b8,b9]	[414, 415]
	422[b8,c2]	120[b8,c2]	140[b8,c2]	356[b8,c2,c3]	221[b8,c2,c3]	0.23[b8,c2]	[414, 415]
	–	–	–	–	316[c4]	–	[417]
	–	–	–	–	356[c5]	–	[417]
	–	–	–	–	320[b8]	–	[418]
	704[b8]	110[b8]	170[b8]	519[b8,c6]	308[b8,c6]	0.22[b8]	[420]
	547[c7]	119[c7]	159[c7]	504[c7,c8]	262[c7,c8]	0.18[c7]	[419]
	370[b8]	110[b8]	153[b8]	320[b8,c9]	197[b8,c9]	0.23[b8]	[419]
	558[b]	146[b]	180[b]	470[b,d1]	305[b,d1]	0.24[b]	[421]
	694[d2]	127[d2]	127[d2]	655[d2,d3]	318[d2,d3]	0.15[d2]	[410]
	–	–	–	–	321[d4]	–	[429, 430]
	–	–	–	–	286[d5]	–	[429]
	–	–	–	–	391[b]	–	[431]
	–	–	–	–	401[d6]	–	[431]
	–	–	–	–	321[d7]	–	[433]
	737[b]	141[b]	175[b]	–	340[b,d8]	–	[434, 435]
	–	–	–	562[d9]	–	–	[541]
	–	–	–	560	–	–	[552]
	–	–	–	290	–	–	[545]
	486[e1]	115[e1]	130[e1]	375[e1,e2]	238[e1,e2]	0.23[e1]	[561, 562]
	778[e3]	128[e3]	181[e3]	563[e3,e4]	344[e3,e4]	0.227[e3]	[617]
	–	–	–	530	–	–	[569]
	–	–	–	–	323	–	[616]
	562[b]	159[b]	146[b]	492[b,b7]	294[b,b7]	0.221[b]	[618]
	–	–	–	525[e5]	–	–	[668]
	–	–	–	466[d3]	283[d3]	–	[694]
	–	–	–	548 ± 26[e6]	–	–	[704]
	747[e7]	113[e7]	175[e7]	544[e7,e8]	325[e7,e8]	0.22[e7]	[716]
	–	–	–	–	311[e9]	–	[716]
	–	–	–	558[f1]	–	–	[721]
	–	–	–	463 ± 43[f2]	–	–	[722, 723]
	–	–	–	458 ± 7[f3]	–	–	[729]
	–	–	–	436 ± 24[f4]	–	–	[771]
	–	–	–	–	310[b,f5]	–	[412]
	632[b]	144[b]	156[b]	466[b,f6]	307[b,f6]	0.25[b]	[844]
	–	–	–	720 (?)	–	–	[338, 397]

[a]For isotropic (or quasi-isotropic) materials [42]: $E = 2G(1 + \nu)$, $E = 3K(1 - 2\nu)$, $K = c_{12} + 2c_{44}/3$, $G = c_{44}$; the condition for the isotropy is given in Eq. (I-2.18)
[b]Calculated on the basis of density-functional theory (DFT) with generalized gradient approximation (GGA) using the Perdew-Burke-Ernzerhof (PBE) functional
[c]Calculated value of G = 147 GPa
[d]Calculated value of G = 195 GPa
[e]Single crystal grown by Verneuil process
[f]Calculated from the reported elastic properties
[g]Estimation from phonon dispersion curves
[h]Hot-pressed in vacuum and annealed, porosity – 7%, mean grain size – 13 μm

(continued)

Table 2.17 (continued)

[i]Sintered in vacuum, porosity – 4.5%, mean grain size – 15 μm
[j]Calculated value of G = 207 GPa
[k]Hot-pressed, content O – 0.01 at.%
[l]Corrected to the poreless state (theoretical density)
[m]Measured experimentally $G = 234 \pm 27$ GPa
[n]Powdered samples
[o]Sintered in vacuum, porosity – 4.0%, mean grain size – 12 μm
[p]Hot-pressed samples, porosity – 6-12%; values measured experimentally (with minimal porosity): E = 505 GPa, G = 239 GPa, K = 294 GPa, $\kappa = 3.33$ TPa^{-1}
[q]Obtained from elastic properties extrapolated to the poreless state
[r]Measured experimentally G = 202 GPa
[s]Calculated from the reported Young's modulus and Poisson's ratio
[t]Measured experimentally G = 216 GPa and $\kappa = 2.84$ TPa^{-1}
[u]Single crystal
[v]Calculated on the basis DFT with GGA
[w]Calculated on the basis of DFT with local density approximation (LDA)
[x]Estimated theoretically by non-conserving pseudopotential and LDA
[y]Nano-powdered samples
[z]Calculated value of G = 169 GPa
[a1]Calculated on the basis of DFT with LDA together with Perdew-Burke-Ernzerhof (PBE) functional
[a2]Calculated value of G = 121 GPa
[a3]Calculated on the basis of DFT with LDA together with the revised Perdew-Burke-Ernzerhof (RPBE) functional
[a4]Calculated value of G = 118 GPa
[a5]Estimation from phonon dispersion curves (c_{11} is extrapolated from neutron data [213])
[a6]Calculated using the Debye–Grüneisen model combined with *ab initio* calculations
[a7]Calculated value of G = 209 GPa
[a8]Calculated value of G = 176 GPa
[a9]Calculated value of G = 188 GPa
[b1]Measured experimentally and corrected for porosity
[b2]$\kappa = 3.24$ TPa^{-1}
[b3]Spark-plasma sintered samples, measured from the segmented micrographs by computing stress analysis and object oriented finite element technique
[b4]Calculated on the basis of density-functional perturbation theory (DFPT)
[b5]From high pressure experiments
[b6]Calculated on the basis of DFT
[b7]Calculated value of G = 166 GPa
[b8]From *ab initio* calculations
[b9]At 0 K
[c1]Calculated value of G = 201 GPa
[c2]At ~300 K
[c3]Calculated value of G = 144 GPa
[c4]Calculated on the basis of DFT with GGA using Perdew's functional
[c5]Calculated on the basis of DFT with GGA using RPBE functional
[c6]Calculated value of G = 213 GPa
[c7]Calculated on the basis of three body force potential (TBFP) model
[c8]Calculated value of G = 181 GPa and $\kappa = 3.82$ TPa^{-1}
[c9]Calculated value of G = 144 GPa and $\kappa = 5.08$ TPa^{-1}
[d1]Calculated value of G = 163 GPa
[d2]Calculated on the basis of interionic potential theory with modified ionic charge
[d3]G = 190 GPa
[d4]Summarized value on the basis of critical reviews of all the available experimental data
[d5]Calculated on the basis of linearized method of "muffin-tin" orbitals (LMTO)
[d6]Calculated on the basis of DFT with LDA together with Ceperley-Alder (CA) type parametrization

(continued)

Table 2.17 (continued)

[d7]Calculated on the basis of DFT with GGA using the plane-wave pseudopotential code Dacapo and Perdew's functional
[d8]Calculated value of G = 217 GPa
[d9]Calculated on the basis of experimental data for some TaC_{1-x} phase containing composites (corrected to zero porosity)
[e1]Calculated on the basis of a modified interaction potential model with covalency ($MIPMC_v$)
[e2]Calculated value of G = 152 GPa
[e3]Calculated on the basis of DFT by projector augmented wave (PAW) method with GGA using PBE functional
[e4]Calculated value of G = 229 GPa
[e5]Experimentally measured by the method of continuous impression of an indenter
[e6]Spark-plasma sintered, porosity – 5.4 ± 0.7%, mean grain size – 4.5 ± 0.9 μm; evaluated through nanoindentation
[e7]Calculated on the basis of DFT using PBE functional with GGA by FP-LAPW method
[e8]G = 223 GPa
[e9]Calculated on the basis of DFT using PBE functional with GGA by plane wave pseudo-potential (PP-PW) method
[f1]Spark-plasma sintered, porosity < 2%, mean grain size – 15-25 μm; measured experimentally by pulse-echo method and extrapolated to zero porosity
[f2]Spark-plasma sintered, no porosity, mean grain size – 5 ± 2 μm; measured experimentally by nanoindentation on the polishing cross-section
[f3]Spark-plasma sintered, porosity < 2%, mean grain size – 5.8 ± 0.2 μm; measured experimentally by Berkovich nanoindentation
[f4]Spark-plasma sintered, porosity – ~7%, mean grain size – 31 ± 2 nm; measured experimentally by indentation technique
[f5]Calculated value of G = 225 GPa
[f6]Calculated value of G = 187 GPa

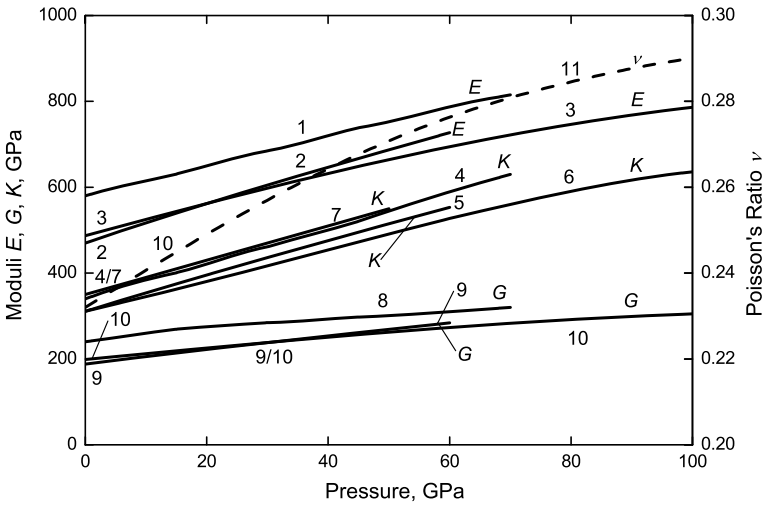

Fig. 2.27 Variations of elastic properties (1-3 – Young's modulus E, 4-7 – bulk (compression) modulus K, 8-10 – Coulomb's (shear) modulus G and 11 – Poisson's ratio v) of near-stoichiometric tantalum monocarbide TaC_{1-x} with higher pressure: 1, 4, 8 – experimentally measured by Chen et al. [200]; 2, 5, 9 – calculated by Peng et al. [204]; 3, 6, 10-11 – calculated by Varshney et al. [414, 415] and 7 – calculated by Gautam and Kumar [617]

Table 2.18 Thermal shock/stress resistance (thermal strength) testing data on single phase tantalum monocarbide TaC_{1-x} materials

Materials and its characterization	Specimen shape and dimensions, mm	Method of thermal loading (testing)	Experimental results	References
$TaC_{\sim 1.0}$, gas-phase deposited, contents: O + N ≤ 0.3%	Hollow cylinders: wall thickness – 0.8–1.2, height – 20	Direct resistance heating with 400 K s^{-1}, cooling by inert gas blow, cycle time – 3 min	Disintegration during the first heating-cooling cycle from 2400 to 50 °C	[41, 163]

c_{44} = 133, c_{12} = 164, c_{13} = 149, c_{14} = –45 (in GPa) [435]. The variations of the elastic moduli E, G and K and Poisson's ratio v with higher pressure (up to 100 GPa) for near-stoichiometric tantalum monocarbide TaC_{1-x} experimentally measured and DFT-calculated are shown in Fig. 2.27. The testing data on thermal shock/stress resistance (thermal strength) of single phase tantalum monocarbide TaC_{1-x} materials are given in Table 2.18.

The magnitudes of physico-mechanical (strength, elasticity) properties of near-stoichiometric tantalum carbides in the wide range of temperatures are summarized in Addendum in comparison with other ultra-high temperature materials.

2.5 Nuclear Physical Properties

The comprehensive lists of isotopes for tantalum Ta and carbon C elements are presented in Tables I-6.2 and I-2.12, respectively. The nuclear physical properties of the elements, including isotopic mass range, total number of isotopes, thermal neutron macroscopic cross sections, moderating ability and capture resonance integral, are given in Table I-A.8.

The thermal neutron macroscopic cross sections for the binary compounds with the molecular formula A_xB_y are calculated according to the following equation:

$$\Sigma_i = [x\sigma_i(\mathrm{A}) + y\sigma_i(\mathrm{B})] \times (N_0\gamma/M), \tag{2.53}$$

where σ_i(A) and σ_i(B) are the microscopic cross sections of the elements A and B, respectively, N_0 is the Avogadro constant, γ is the density and M is the molecular mass of the compound. The calculated cross sections of near-stoichiometric tantalum carbide phases are listed in Table 2.19.

Table 2.19 Thermal neutron macroscopic cross sections[a] of near-stoichiometric tantalum carbides [16, 217]

Phase	Macroscopic thermal neutron cross section of capture (absorption) Σ_a, cm^{-1}	Macroscopic thermal neutron cross section of scattering Σ_s, cm^{-1}
α-$Ta_{2+x}C$	1.00	~0.41
TaC_{1-x}	0.93	~0.48

[a]For 2200 m s^{-1} neutrons

Table 2.20 Parameters of formation and migration of lattice point defects (vacancies and interstitial atoms) in near-stoichiometric tantalum monocarbide[a]

Defect	Metal sublattice				Non-metal sublattice			
	E_f, eV	E_m, eV	S_f/k_B	S_m/k_B	E_f, eV	E_m, eV	S_f/k_B	S_m/k_B
Vacancy[b] [218, 739]	4.25	3.54	4.59	3.83	1.42	5.68	1.53	6.12
Vacancy[c] [538, 739]	3.79	3.31	4.11	3.60	1.10	3.31	1.19	3.60
Vacancy[d] [434]	3.5	–	–	–	0.26	4.0[e]	–	–
Vacancy[f] [735]	1.63	–	–	–	–	–	–	–
Interstitial atom[b] [218]	20.59	0.71	22.37	0.77	6.39	0.71	6.94	0.77

[a]Denoted: E_f – defect formation energy, E_m – defect migration energy, S_f – defect formation entropy, S_m – defect migration entropy, k_B – Boltzmann constant

[b]Calculated on the basis of bonding model (relaxation displacement of atoms surrounding a defect is not taking into account)

[c]Calculated on the basis of elastic continuum model

[d]*Ab initio* simulation computed from the largest super-cell on the basis of the PAW pseudopotential using the PBE formulation

[e]Calculated value of activation energy for carbon self-diffusion in tantalum monocarbide E_A = 415 kJ mol^{-1} (*see also* Table 2.25)

[f]Calculated on the basis of data on the yield strength determined by hardness and elasticity experimental measurements

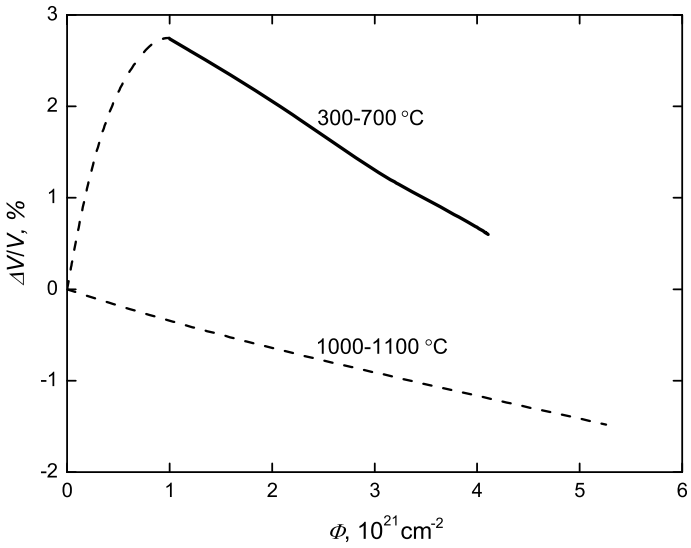

Fig. 2.28 The change in macroscopic volume (swelling/shrinkage) with fast neutron (>1 MeV) fluence Φ for produced by ceramic technologies (hot-pressed, slip-cast, explosion-pressed) near-stoichiometric tantalum monocarbide TaC_{1-x} materials irradiated at different temperatures [218, 219, 619, 845, 846]

For the estimation of probable damage of tantalum monocarbide materials exposed to various types of radiation the parameters of formation and migration of lattice point defects (vacancies and interstitial atoms) are given in Table 2.20. Examples of such damage (swelling/shrinkage) produced by fast neutrons (>1 MeV) in the ranges of moderate and elevated temperatures are shown in Fig. 2.28. The fast neutron fluence more than 2×10^{21} cm^{-2} results in intensive failure of near-stoichiometric tantalum monocarbide hot-pressed (or sintered) materials [219]. As a result of damage the radiation-induced resistivity in $TaC_{0.99}$ increases noticeably but less than the same parameter in other refractory carbides; the contribution of the different defects to the damage rates were tentatively separated [220, 740]. The damage rate and recovery processes in $TaC_{0.99}$ irradiated with electrons of different energies were evaluated by Gosset et al. [221] and Allison et al. [331]; the threshold energy for carbon displacement was determined to be 23.2 ± 1.1 eV. The damage produced in TaC_{1-x} coating by bombardment with 90 keV protons to a dose 1.24×10^{21} cm^{-2} comprises erosion of the materials and its sputter etching; blisters could not be detected [401]. Preferential sputtering of carbon from the surface of TaC_{1-x} (depletion of carbon constituent) due to light ion (He^+, H_2^+) bombardment ($E = 0.3$–4 keV, $\Phi = \sim 10^{18}$ cm^{-2}) was observed by Varga and Taglauer [847]. It was found that no significant deformations occur with deuterium ions D^+ implantation up to $\sim 10^{18}$ cm^{-2} at 800 °C on tantalum monocarbide; however, steady-state irradiation in the range of temperatures 750–1250 °C with energetic (20–40 keV) He ions fluences between 6×10^{17} and 1×10^{20} cm^{-2} leads to extensive pore formation [222–224]. Deuterium retention in toughened, fine-grain and recrystallized W doped with 3.2 mol.% TaC_{1-x}, irradiated by 38 eV D^+ ions to fluences in the range of $6.0 \times 10^{20} - 1.8 \times 10^{21}$ cm^{-2}, was studied at 525 °C by Zibrov et al. [748]: the deuterium inventory in the materials was very low and comparable with that in pure polycrystalline W. Employing a positron annihilation method for the studies of atomic defects in the irradiated transition metal monocarbides at low temperatures, Rempel et al. [740, 848] proved that the recombination of interstitial tantalum atoms and vacancies in $TaC_{0.99}$ occurred at temperatures below 85 K.

Nuclear physical properties of tantalum carbides in comparison with other ultra-high temperature materials are also given in Addendum.

2.6 Chemical Properties and Materials Design

The comprehensive data on the chemical properties, compatibility (in the connection with both environmental resistance and composite materials design) and interaction behaviour of tantalum carbide phases at elevated, high and ultra-high temperatures with elements (metals, non-metals) are summarized in Table 2.21, with refractory compounds – in Table 2.22 and with gaseous media – in Table 2.23. The data on the oxidation resistance of tantalum carbide materials listed there are also accompanied by the graphic information in Figs. 2.29 and 2.30; the isothermal

Table 2.21 Chemical interaction and/or compatibility of tantalum carbide phases with elements (metals, non-metals) at elevated, high and ultra-high temperatures, including solid matters and molten media (reaction systems are given mainly in alphabetical order)[a]

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	References
TaC_{1-x}–Ag	–	1700	Addition of 5 mol.% Ag intensifies the densification of powdered carbide by spark plasma sintering (pressure 30 MPa, exposure 5 min)	[302]
TaC_{1-x}–Al	–	<1000	No interaction, lattice parameter of TaC_{1-x} is invariable; the particles of TaC_{1-x} show high stability and uniform distribution in melts	[1, 16, 225, 273–275, 302, 513, 514, 689–692, 742, 794, 876–878]
	He	1000–1400	Formation of Al_4C_3; at higher temp. it decomposes	
	–	>1250	The formation of solid solutions; lattice parameter of TaC_{1-x} is decreasing considerably	
	–	–	Formation of Ta_2AlC, Ta_3AlC_2, $Ta_5Al_3C_{1-x}$, Ta_3Al_2C and α/β-Ta_4AlC_3 ($M_{n+1}AX_n$-phases)	
	–	1400–2800	Addition of 1% Al does not affect the densification of powdered carbide by hot-pressing	
	–	1700	Testing in Al vapours results in the formation of refractory phase impregnated with Al in pores	
	–	1700	Addition of 5 mol.% Al intensifies the densification of powdered carbide by SPS (pressure 30 MPa, exposure 5 min) *See also* section TaC_{1-x}–Al_4C_3 in Table 2.22 *See also* Table 2.24 *See also* section C–Al–Ta in Table I-2.14	
TaC_{1-x}–Al–Co	Ar	1350	Formation of η-$Ta_3(Co,Al)_3C_{1\pm x}$ phases	[879]
TaC_{1-x}–Al–Cr	Ar	1350	Formation of η-$Ta_3(Cr,Al)_3C_{1\pm x}$ phases	[879]
TaC_{1-x}–Al–Cu	Ar	1350	Formation of η-$Ta_3(Cu,Al)_3C_{1\pm x}$ phases	[879]
TaC_{1-x}–Al–Fe	–	1000	Particles of TaC_{1-x} are rather stable in Fe – 15-26 at.% Al alloys during long-term exposures, although the formation of ε-$(Fe,Al)_{2\pm x}Ta$ is observed; the solubility of Fe in carbide varies from 0.9 to 5.0 at.% and that of Al is very low	[822, 823, 879]
	Ar	1350	Formation of η-$Ta_3(Fe,Al)_3C_{1\pm x}$ phases	
TaC_{1-x}–Al–Mn	Ar	1350	Formation of η-$Ta_3(Mn,Al)_3C_{1\pm x}$ phases	[879]
TaC_{1-x}–Al–Ni	Ar	1350	Formation of η-$Ta_3(Ni,Al)_3C_{1\pm x}$ phases	[879]
TaC_{1-x}–Al–Sn	–	–	Formation of $Ta_3(Al,Sn)C_2$ ($M_{n+1}AX_n$-phase); $Ta_3(Al_{0.6}Sn_{0.4})C_2$ was synthesized	[708, 709]
TaC_{1-x}–Al–V	Ar	1350	Formation of η-$Ta_3(V,Al)_3C_{1\pm x}$ phases	[879]
TaC_{1-x}–Al–Zn	Ar	1350	Formation of η-$Ta_3(Zn,Al)_3C_{1\pm x}$ phases	[879]
TaC_{1-x}–TiC_{1-x}–Al	–	–	Formation of $(Ta,Ti)_2AlC$ ($M_{n+1}AX_n$-phase solid solution); $(Ta_{0.6}Ti_{0.4})_2AlC$ was synthesized *See also* section C–Al–Ta–Ti in Table I-2.14	[273–275, 469, 708]

(continued)

Table 2.21 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	References
TaC_{1-x}–VC_{1-x}–Al	–	–	Formation of $(Ta,V)_2AlC$ ($M_{n+1}AX_n$-phase solid solution); $(Ta_{0.35}V_{0.65})_2AlC$ was synthesized *See also* section C–Al–Ta–V in Table I-2.14	[273–275, 469, 708]
α-$Ta_{2+x}C$–Al			*See* section C–Al–Ta in Table I-2.14	
TaC_{1-x}–Au	–	1700	Addition of 5 mol.% Au intensifies the densification of powdered carbide by SPS (pressure 30 MPa, exposure 5 min) considerably	[302]
TaC_{1-x}–β-B	–	>2400	Formation of $TaB_{2\pm x}$ *See also* section C–B–Ta in Table I-2.14	[1]
α-$Ta_{2+x}C$–β-B			*See* section C–B–Ta in Table I-2.14	
TaC_{1-x}–Be	–	1400–2800	Addition of 1% Be does not affect the densification of powdered carbide by hot-pressing	[16, 343, 772]
	Vacuum	1500	Formation of terminal monocarbide (cubic) solid solution of substitutional-interstitial type	
	–	–	Formation of $Be_xTaC_{0.76}$ (?) *See also* section C–Be–Ta in Table I-2.14	
TaC_{1-x}–Bi			*See* Table 2.24	
TaC_{1-x}–C	Ar/He	1900–2400	Addition of 0.1–0.8% C intensifies the densification of powdered carbide by hot-pressing	[1, 4, 5, 16, 61, 188, 301, 463–468, 675, 697, 714]
	Vacuum	<3300	TaC_{1-x} is compatible in contact with C (graphite) parts	
	–	~3400–3450	Eutectic TaC_{1-x}–C (graphite) *See also* section C–Ta in Table I-2.13	
TaC_{1-x}–C–Fe			*See* Table 2.24	
TaC_{1-x}– C–Fe –Mn			*See* Table 2.24	
TaC_{1-x}–β-$Mo_{2\pm x}C$–TiC_{1-x}–δ-$TiN_{1\pm x}$ –δ-$WC_{1\pm x}$–Ce–Co–Ni	–	1470–1500	Formation of complex monocarbonitride $(Ta,Ti,W,Mo)(C,N)_{1-x}$ and complex intermetallide $(Ti,W,Mo,Ce)(Co,Ni)_y$ with core-rim microstructures	[803]

(continued)

Table 2.21 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	References
TaC_{1-x}–α/ε–Co	–	1100	TaC_{1-x} is in equilibrium with λ_2-$Co_{2\pm x}Ta$ and α-Co; η-$Ta_4Co_2C_{1-x}$ is formed	[16, 226, 265, 270, 271, 293, 457, 460, 784, 815, 871, 881]
	–	~1250	The max. solid solubility of TaC_{1-x} in α-Co is ~1.0–2.5 mol.%	
	–	1330–1400	Metallic Co on TaC_{1-x} plate instantaneously penetrates into the carbide and forms a eutectic (the process is identical in the case of both sintered and fused carbide materials)	
	–	1340–1370	Eutectic TaC_{1-x}–α-Co; the solubility of TaC_{1-x} in α-Co is ~0.5–1.0 mol.%	
	–	1400	The max. solubility of TaC_{1-x} in molten Co is ~2.0 mol.%	
	–	1450	The solubility of Ta in Co-based binder phase in equilibrium with C is 4.0–4.5%	
	–	1500	The max. solubility of TaC_{1-x} in molten Co is ~4.0 mol.%	
	–	2750	Addition of 0.5–2.5 mas.% Co intensifies the densification of powdered carbide by hot-pressing *See also* Table 2.24 *See also* section C–Co–Ta in Table I-2.14	
TaC_{1-x}–Cr_3C_{2-x}–β-$Mo_{2\pm x}C$–TiC_{1-x}–δ-$TiN_{1\pm x}$–δ-$WC_{1\pm x}$–Co–Ni	Vacuum	1400–1410	Formation of $(Ta,Ti,W,Mo,Cr)(C,N)_{1-x}$–$Ti(C,N)_{1-x}$–(Co,Ni,Mo) quasi-three-phase system with core-rim microstructures	[804]
TaC_{1-x}–Cr_3C_{2-x}–NbC_{1-x}–TiC_{1-x}–Co	–	–	Formation of $(Ta,Ti,Nb,Cr)C_{1-x}$–Co hard alloys (cermets)	[849]
TaC_{1-x}–HfC_{1-x}–Co	–	~1420–1445	Eutectic $(Ta,Hf)C_{1-x}$ (x = 0.06–0.08)–α-Co *See also* section C–Co–Hf–Ta in Table I-2.14	[270]
TaC_{1-x}–HfC_{1-x}–β-$Mo_{2\pm x}C$–TiC_{1-x}–δ-$TiN_{1\pm x}$–δ-$WC_{1\pm x}$–Co	Vacuum	1400–1500	Formation of $(Ta,Hf,Ti,W,Mo)(C,N)_{1-x}$–Co two-phase cermet system with core-rim structures	[781]
TaC_{1-x}–β-$Mo_{2\pm x}C$–TiC_{1-x}–δ-$TiN_{1\pm x}$–VC_{1-x}–δ-$WC_{1\pm x}$–Co–Ni	Ar	1400	Formation of $(Ta,Ti,W,Mo,V)(C,N)_{1-x}$–$Ti(C,N)_{1-x}$–(Co,Ni,Mo) quasi-three-phase system with core-rim microstructures	[860]
TaC_{1-x}–β-$Mo_{2\pm x}C$–TiC_{1-x}–δ-$TiN_{1\pm x}$–δ-$WC_{1\pm x}$–Co–Ni	–	1400–1600	Formation of $(Ta,Ti,W,Mo)(C,N)_{1-x}$–(Co,Ni,Mo) cermet system with core-rim microstructures	[780, 790–792, 798, 857, 859–861]

(continued)

Table 2.21 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	References
TaC_{1-x}–β-$Mo_{2\pm x}C$–TiC_{1-x}–δ-$WC_{1\pm x}$–Co–Ni	–	1460	Formation of $(Ta,Ti,W,Mo)C_{1-x}$–(Co,Ni) two-phase cermet system with core-rim structure (the core structure of mixed carbide grains with either a high W or Ti content and rim structure with different metallic elements and distributions)	[785]
TaC_{1-x}–NbC_{1-x}–Co	–	~1410–1445	Eutectic $(Ta,Nb)C_{1-x}$ (x = 0.18)–α-Co *See also* section C–Co–Nb–Ta in Table I-2.14	[270, 775]
TaC_{1-x}–NbC_{1-x}–TiC_{1-x}–δ-$TiN_{1\pm x}$–Co	Pure He	1550	Formation of complex monocarbonitride $(Ta,Nb,Ti)(C,N)_{1-x}$ and complex intermetallides $(Ti,Ta,Nb)Co_{1\pm x}$, α-$(Ti,Ta,Nb)Co_{2\pm x}$ and β-$(Ti,Ta,Nb)Co_{2+x}$	[783]
TaC_{1-x}–NbC_{1-x}–TiC_{1-x}–δ-$TiN_{1\pm x}$–Co–Ni	Pure He	1550	Formation of complex monocarbonitride $(Ta,Nb,Ti)(C,N)_{1-x}$ and complex intermetallide $(Ti,Ta,Nb)(Ni,Co)_{3\pm x}$ (orthorhombic modification)	[783, 864]
TaC_{1-x}–NbC_{1-x}–TiC_{1-x}–δ-$TiN_{1\pm x}$–VC_{1-x}–δ-$WC_{1\pm x}$–Co–Mo–Ni	–	–	Formation of complex monocarbonitride (cubic) $(Ta,Nb,Ti,V,W)(C,N)_{1-x}$ phase	[793]
TaC_{1-x}–NbC_{1-x}–TiC_{1-x}–δ-$TiN_{1\pm x}$ – δ-$WC_{1\pm x}$–Co	–	–	General consideration of the $(Nb,Ta,Ti,W)(C,N)_{1-x}$–δ-$(W,Nb)C_{1\pm x}$–α-Co system	[746, 863, 864]
TaC_{1-x}–NbC_{1-x}–TiC_{1-x}–δ-$WC_{1\pm x}$–Co	–	–	General consideration of the $(Nb,Ta,Ti,W)C_{1-x}$–δ-$(W,Nb)C_{1\pm x}$–α-Co system is presented	[272, 746, 784, 869, 872]
TaC_{1-x}–NbC_{1-x}–δ-$WC_{1\pm x}$–Co	–	–	General consideration of the $(Nb,Ta,W)C_{1-x}$–δ-$(W,Nb)C_{1\pm x}$–α-Co system is presented	[872]
TaC_{1-x}–TiC_{1-x}–Co	– –	~1390–1450 1500	Eutectic $(Ta,Ti)C_{1-x}$ (x = 0.18–0.20)–α-Co Formation of $(Ta,Ti)C_{1-x}$–Co two-phase cermet system *See also* section C–Co–Ta–Ti in Table I-2.14	[270, 850]
TaC_{1-x}–TiC_{1-x}–δ-$TiN_{1\pm x}$–Co	Pure Ar/N_2, vacuum	1400–1600	Formation of complex monocarbonitride $(Ta,Ti)(C,N)_{1-x}$ and complex intermetallides (in low concentrations) $(Ti,Ta)Co_{1\pm x}$, α-$(Ti,Ta)Co_{2\pm x}$ and β-$(Ti,Ta)Co_{2+x}$	[782, 797, 873]
TaC_{1-x}–TiC_{1-x}–δ-$TiN_{1\pm x}$–δ-$WC_{1\pm x}$–Co–Ni	Ar	1550	The liquid-phase sintering of powdered $TiC_{0.5}N_{0.5}$ – 1.5 mol.% TaC_{1-x} – 1.5 mol.% δ-$WC_{1\pm x}$ – 11 mol.% Co – 11 mol.% Ni composition leads to the formation of $Ti(C,N)_{1-x}$ core/$(Ti,W,Ta)(C,N)_{1-x}$ rim – microstructured hard alloy (cermet) system	[865]

(continued)

Table 2.21 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	References
TaC_{1-x}–TiC_{1-x}–δ-$WC_{1\pm x}$–Co	–	1250–1450	The max. solubility of (Ta,Ti,W)C_{1-x} in Co increases with temperature growth from ~3 to ~5 mol.%	[790, 850, 851, 867, 868, 870, 885]
	Vacuum	1450	Formation of (Ta,Ti,W)C_{1-x} – 30% Co two-phase and (Ta,Ti,W)C_{1-x}–δ-$WC_{1\pm x}$ – 30% Co three-phase hard alloy (cermet) systems with Ti and W contents in the metal binder: 0.2% and 8% (two-phase system) and 0.1% and 10% (three-phase system), respectively; decrease in C content in the hard phase (with constant 10% molar fraction of TaC_{1-x} in it) from (Ta,Ti,W)$C_{1.0}$ to (Ta,Ti,W)$C_{0.85}$ leads to the increase of Ti and W contents in the metal binder – up to 1.2% and 22%, respectively	
	–	1500	Formation of (Ta,Ti,W)C_{1-x}–Co two-phase hard metal (cermet) system with core-rim microstructures (with high Ti and Ta contents)	
TaC_{1-x}–TiC_{1-x}–δ-$TiN_{1\pm x}$–δ-$WC_{1\pm x}$–Co–Ni	Vacuum	1500	Formation of (Ta,Ti,W)$(C,N)_{1-x}$–(Co,Ni) two-phase cermet system with core-rim microstructures	[790, 855, 856]
TaC_{1-x}–VC_{1-x}–Co	–	~1410–1445	Eutectic (Ta,V)C_{1-x} (x = 0.18)–(V,Ta)C_{1-x} (x = 0.17)–α-Co *See also* section C–Co–Ta–V in Table I-2.14	[271]
TaC_{1-x}–δ-$WC_{1\pm x}$–Co	–	–	General consideration of the (Ta,W)C_{1-x}–δ-(W,Ta)$C_{1\pm x}$–α-Co system are presented	[272, 680, 765, 790, 866]
	–	1400	The solid solubility of Ta in η_2-$(Co,W)_6C$ is corresponding to $Co_{3.04}W_{2.74}Ta_{0.22}C$ composition *See also* section C–Co–Ta–W in Table I-2.14	
TaC_{1-x}–ZrC_{1-x}–Co	–	1250	Eutectic (Ta,Zr)C_{1-x} (x = 0.18)–(Zr,Ta)C_{1-x} (x = 0.19)–α-Co *See also* section C–Co–Ta–Zr in Table I-2.14	[271]
α-$Ta_{2+x}C$–Co			*See* section C–Co–Ta in Table I-2.14	
TaC_{1-x}–Cr	–	~1675–1695	Eutectic TaC_{1-x}–Cr; the solubility of Cr in TaC_{1-x} is ~3.4 mol.% and that of TaC_{1-x} in Cr is ~2 mol.%	[226–229, 456, 681]
	–	~1695	Eutectic (Ta,Cr)C_{1-x} (x = 0.21)–(Cr_yTa_z) (y = 0.991, z = 0.009) *See also* section C–Cr–Ta in Table I-2.14	
α-$Ta_{2+x}C$–Cr	–	–	The solubility of Cr in α-$Ta_{2+x}C$ is <2 at.% *See also* section C–Cr–Ta in Table I-2.14	[123, 226]

(continued)

Table 2.21 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	References
TaC_{1-x}–Cs	Cs vapour, ~13 Pa	~2000	No effect upon structural characteristics of TaC_{1-x} and its thermionic emission parameters (some tens of hours exposure)	[147, 149, 458]
α/β-$Ta_{2+x}C$–Cs	Cs vapour	–	No irreversible phenomena observed	[445, 458]
TaC_{1-x}–Cu	Vacuum	~1200	No noticeable interaction between carbide particles (mean size – 1-2 μm) and Cu melt	[16, 302, 763]
	–	1700	Addition of 5 mol.% Cu intensifies the densification of powdered carbide by SPS (pressure 30 MPa, exposure 5 min) considerably *See also* Table 2.24	
TaC_{1-x}–Cu–Zr	–	1300	The infiltration of TaC_{1-x} preforms (porosity – 55%) by Cu – 22-67 at.% Zr alloy leads to the formation of compositions containing ZrC_{1-x} ($x \approx 0.4$) and α-$Ta_{2+x}C$ phases along with the initial TaC_{1-x} and Cu constituents	[880]
TaC_{1-x}–$\alpha/\gamma/\delta$-Fe	–	1000	The interaction initiates along the grain boundaries of bulk γ-Fe	[16, 226, 265, 292, 457, 460, 628, 815, 874]
	–	1100	TaC_{1-x} ($x > 0.15$) is in equilibrium with ε-$Fe_{2\pm x}Ta$ and TaC_{1-x} ($0.01 \leq x \leq 0.15$) – with γ-Fe	
	–	1200	Formation of new solid phases (20 μm layer) in the contact zone between the bulk components	
	–	1250	The max. solubility of TaC_{1-x} in γ-Fe is ~0.15–0.30 mol.%	
	–	1360–1440	Metallic Fe on TaC_{1-x} plate instantaneously penetrates into the carbide and forms a eutectic (the process is identical in the case of both sintered and fused carbide materials)	
	–	~1440	Eutectic TaC_{1-x}–δ-Fe	
	–	2750	Addition of 0.5–2.5% Fe intensifies the densification of powdered carbide by hot-pressing	
	–	–	DFT calculated semicoherent interfacial energy at relaxed interface – 0.38 J m^{-2} *See also* Table 2.24 *See also* section C–Fe–Ta in Table I-2.14	
TaC_{1-x}–TiC_{1-x}–δ-$TiN_{1\pm x}$–Fe α-$Ta_{2+x}C$–Fe	Pure N_2, vacuum	1500	Formation of $(Ta,Ti)(C,N)_{1-x}$–Fe two-phase cermet system *See* section C–Fe–Ta in Table I-2.14	[797]
TaC_{1-x}–Ga	–	–	Formation of Ta_2GaC ($M_{n+1}AX_n$-phase) *See also* Table 2.24 *See also* section C–Ga–Ta in Table I-2.14	[273–275, 461]

(continued)

Table 2.21 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	References
TaC_{1-x}–Ge			*See* Table 2.24	
TaC_{1-x}–Hf			*See* section C–Hf–Ta in Table I-2.14	
α-$Ta_{2+x}C$–Hf			*See* section C–Hf–Ta in Table I-2.14	
TaC_{1-x}–In			*See* Table 2.24	
TaC_{1-x}–Ir	–	~2030–2130	Eutectic (estimated) TaC_{1-x} ($x \approx 0.18$)–Ir *See also* section C–Ir–Ta in Table I-2.14	[226, 344]
α-$Ta_{2+x}C$–Ir			*See* section C–Ir–Ta in Table I-2.14	
TaC_{1-x}–K	–	~2000	No chemical reaction observed, no effect upon structural characteristics of TaC_{1-x} and its thermionic emission parameters (some tens of hours exposure)	[149, 458]
TaC_{1-x}–α/β/γ/δ-Mn	–	–	The effect of Mn as an additive for hot-pressing process was studied *See also* Table 2.24	[16, 815]
TaC_{1-x}–Mo	–	1400	Under pressure 5 MPa the diffusion welded joint between TaC_{1-x} and Mo is produced (5–15 min exposure)	[1, 16, 21, 61, 215, 283, 285, 286, 462, 801, 802]
	Vacuum	≥1600	The initiation of reaction (5 h exposure) between powdered TaC_{1-x} and compact dense metallic Mo	
	Vacuum	2100–2200	Weak contact interaction (5 h exposure) between the bulk components and formation of mixed carbide phases *See also* section C–Mo–Ta in Table I-2.14	
α-$Ta_{2+x}C$–Mo			*See* section C–Mo–Ta in Table I-2.14	
TaC_{1-x}–Cr_3C_{2-x}–TiC_{1-x}–δ-$WC_{1\pm x}$–Mo–Ni	Vacuum, 0.01–0.1 Pa	1450	The liquid-phase sintering of powdered $(Ta_{0.14}Ti_{0.65}W_{0.21})C_{1-x}$–$Cr_3C_{2-x}$–Mo–Ni compositions leads to the formation of core-rim structured (Ta,Ti,W,Mo)C_{1-x} phases embedded in the metallic Ni-based binder	[862]
TaC_{1-x}–TiC_{1-x}–δ-$TiN_{1\pm x}$–δ-$WC_{1\pm x}$–ZrC_{1-x}–Mo–Ni	–	–	The formation of two- or three-phase cermet systems	[885]
TaC_{1-x}–TiC_{1-x}–Mo–Ni	Vacuum, 0.7–1.3 Pa	1200–1400	Hard alloys $(Ti_{0.74–0.98}Ta_{0.02–0.26})C_{0.93–0.97}$ – 12% (Ni,Mo) were prepared	[883, 884]
TaC_{1-x}–TiC_{1-x}–δ-$WC_{1\pm x}$–Mo–Ni	–	1400–1500	Formation of (Ta,Ti,W,Mo)C_{1-x}–(Ni,Mo) hard alloy (cermet) system	[850, 885]

(continued)

Table 2.21 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	References
TaC_{1-x}–Mo–Re	Vacuum	2450	No interaction in the contact zone (1 h exposure) between Mo – 40 mas.% Re alloy and compact dense carbide materials	[287]
	Vacuum or He	2500	Interaction in the contact zone (15 min. exposure) between the compact dense materials	
	Vacuum or He	>2625	Being absorbed into the pores molten alloy dissolves (~25 vol.%) in TaC_{1-x} (25% porosity) materials completely *See also* sections C–Mo–Ta and C–Re–Ta in Table I-2.14	
TaC_{1-x}–Nb	Vacuum	1200	The initiation of reaction in powdered mixtures with the formation of solid solutions; under pressure 5 MPa the diffusion welded joint between TaC_{1-x} and Nb parts is produced (5–15 min exposure)	[1, 16, 37, 61, 215, 283–285]
	Vacuum	≥1600	The initiation of reaction between powdered TaC_{1-x} and compact dense Nb (exposure 5 h)	
	Vacuum	1600–1800	Formation of mixed carbide phases (α-$Ta_{2+x}C$ –β-$Nb_{2+x}C$ and TaC_{1-x}–NbC_{1-x} solid solutions)	
	Vacuum	>2000–2200	Intensive interaction between carbide and metallic phases *See also* section C–Nb–Ta in Table I-2.14	
α-$Ta_{2+x}C$–Nb			*See* section C–Nb–Ta in Table I-2.14	
TaC_{1-x}–Ni	Vacuum	1000	Formation of new solid phases initiates in the contact zone	[16, 226, 239, 265, 292, 457, 784, 815, 858, 881]
	–	1100	TaC_{1-x} is in equilibrium with Ni-Ta intermetallides and metal Ni; η-$Ta_4Ni_2C_{1-x}$ is formed	
	–	1200	12 μm layer of new solid phases forms in the contact zone for 1 h exposure	
	–	1250	The max. solubility of TaC_{1-x} in metallic Ni is ~1.5–3.0 mol.%	
	–	1300–1350	Metallic Ni on TaC_{1-x} plate instantaneously penetrates into the carbide and forms a eutectic (the process is identical in the case of both sintered and fused carbide materials)	
	–	~1340–1370	Eutectic TaC_{1-x}–Ni; the max. solubility of TaC_{1-x} in solid metallic Ni is ~2–4 mol.%	
	–	~1400	The max. solubility of TaC_{1-x} in molten Ni is 2.0 mol.%	
	–	2750	Addition of 0.5–2.5% Ni intensifies the densification of powdered carbide by hot-pressing *See also* Table 2.24 *See also* section C–Ni–Ta in Table I-2.14	
TaC_{1-x}–Ni–Cr	–	–	TaC_{1-x} (porosity – 6.0%) is impregnated perfectly by the Ni – 9.5 mas.% Cr alloy *See also* sections C–Ni–Ta and C–Cr–Ta in Table I-2.14	[1]

(continued)

Table 2.21 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	References
TaC_{1-x}–HfC_{1-x}–β-$Mo_{2\pm x}C$–TiC_{1-x}–δ-$TiN_{1\pm x}$–δ-$WC_{1\pm x}$–Ni	Vacuum	1400–1500	Formation of $(Ta,Hf,Ti,W,Mo)(C,N)_{1-x}$–(Ni,Mo) two-phase hard alloy (cermet) system with core-rim microstructures	[781]
TaC_{1-x}–HfC_{1-x}–TiC_{1-x}–δ-$TiN_{1\pm x}$–Ni	Vacuum	1500	The interaction of $(Ta,Hf,Ti)C_{0.70}N_{0.30}$ monocarbonitride phase with Ni is considered	[786]
TaC_{1-x}–β-$Mo_{2\pm x}C$–TiC_{1-x}–Ni	–	1350–1410	Powdered TiC_{1-x} – 3.5 mol.% β-$Mo_{2\pm x}C$ – 3.5 mol.% TaC_{1-x} – 23.5 mol.% Ni compositions were sintered to form hard alloys (cermets)	[853]
TaC_{1-x}–β-$Mo_{2\pm x}C$–TiC_{1-x}–δ-$TiN_{1\pm x}$–Ni	–	1350–1480	Powdered Ti(C,N) – 2-22.5 vol.% TaC_{1-x} – 7 vol.% β-$Mo_{2\pm x}C$ – 8 vol.% Ni compositions were heat-treated to form cermets	[852, 854]
TaC_{1-x}–β-$Mo_{2\pm x}C$–TiC_{1-x}–δ-$TiN_{1\pm x}$–δ-$WC_{1\pm x}$–Ni	–	1600	Formation of $(Ta,Ti,W,Mo)(C,N)_{1-x}$ – Ni two-phase cermet system with core-rim structures	[780]
TaC_{1-x}–NbC_{1-x}–Ni	–	~1330–1350	Eutectic $(Ta,Nb)C_{1-x}$ (x = 0.1–0.2) – Ni; the max. solubility of $(Nb,Ta)C_{1-x}$ in Ni is ~3.0-3.5 mol.% *See also* section C–Nb–Ni–Ta in Table I-2.14	[277, 775]
TaC_{1-x}–NbC_{1-x}–TiC_{1-x}–δ-$TiN_{1\pm x}$–Ni	Pure He	1550	Formation of complex monocarbonitride $(Ta,Nb,Ti)(C,N)_{1-x}$ and complex intermetallide $(Ti,Ta,Nb)Ni_{3\pm x}$ (hexagonal and orthorhombic modifications)	[783]
TaC_{1-x}–TiC_{1-x}–Ni	–	1000	Phase diagram of the system possesses one monovariant eutectic, but the alloys crystallize in a metastable state, as the carbide phase does not decompose into 2 solid solutions even after long-term annealing (10 h exposure)	[738]
TaC_{1-x}–TiC_{1-x}–δ-$TiN_{1\pm x}$–Ni	Vacuum	~1400–1500	The interaction of $(Ta,Ti)C_{0.50–0.70}N_{0.30–0.50}$ carbonitride phases with Ni is considered	[747, 773, 776–778, 786, 797]
TaC_{1-x}–TiC_{1-x}–δ-$TiN_{1\pm x}$–δ-$WC_{1\pm x}$–Ni	Vacuum	1500	The interaction of $(Ta,Ti,W)C_{0.70}N_{0.30}$ monocarbonitride phase with metallic Ni is considered	[786, 865]
	Ar	1550	The liquid-phase sintering of powdered $TiC_{0.5}N_{0.5}$ – 1.5 mol.% TaC_{1-x} – 1.5 mol.% δ-$WC_{1\pm x}$ – 22 mol.% Ni composition leads to the formation of $Ti(C,N)_{1-x}$ core/$(Ti,W,Ta)(C,N)_{1-x}$ rim – microstructured hard alloy (cermet) system	
TaC_{1-x}–ZrC_{1-x}–Ni	–	1000	Max. total solid solubility of carbides in Ni is $\leq$ ~2–3 mol.%	[807, 808]
	–	~1240–1270	Eutectic $(Ta,Zr)C_{1-x}$ (x = 0.02)–$(Zr,Ta)C_{1-x}$ (x = 0.17)–Ni	

(continued)

Table 2.21 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	References
α-$Ta_{2+x}C$–Ni			*See* section C–Ni–Ta in Table I-2.14	
TaC_{1-x}–Os	–	1500	The solid solubility of Os in TaC_{1-x} is low *See also* section C–Os–Ta in Table I-2.14	[226]
α-$Ta_{2+x}C$–Os			*See* section C–Os–Ta in Table I-2.14	
TaC_{1-x}–Pb			*See* Table 2.24	
TaC_{1-x}–Pd	–	1200–1500	The formation of (Pd,Ta) solid solutions and failure of TaC_{1-x} are occurred (8 h exposure); TaC_{1-x} is in equilibrium with ($Pd_{0.84}Ta_{0.16}$) binary alloy *See also* section C–Pd–Ta in Table I-2.14	[1, 16, 226, 444]
α-$Ta_{2+x}C$–Pd			*See* section C–Pd–Ta in Table I-2.14	
TaC_{1-x}–Pt	–	1200–1500	The formation of (Pt,Ta) solid solutions and failure of TaC_{1-x} are occurred (8 h exposure); TaC_{1-x} is in equilibrium with $Pt_{3\pm x}Ta$ *See also* section C–Pt–Ta in Table I-2.14	[1, 16, 226, 444]
α-$Ta_{2+x}C$–Pt			*See* section C–Pt–Ta in Table I-2.14	
TaC_{1-x}–Pu			*See* section C–Pu–Ta in Table I-2.14	
α-$Ta_{2+x}C$–Pu			*See* section C–Pu–Ta in Table I-2.14	
TaC_{1-x}–Re	–	~2420–2450	Eutectic TaC_{1-x}–Re; the max. solid solubility of Re in TaC_{1-x} is ~3–4 mol.% and that of TaC_{1-x} in Re is low too	[226, 243, 244, 287–290, 745]
	Vacuum	2500	No interaction in the contact zone (1 h exposure) between the compact dense materials	
	–	–	Stable and compatible with each other as cermet components	
	He	2655 ± 25	Melting 80 mol.% $TaC_{0.94}$ + 20 mol.% Re composition results in the formation of $TaC_{0.90}$ + Re-based solid solution	
	He	2700 ± 25	Melting 20 mol.% $TaC_{0.70}$ + 80 mol.% Re composition results in the formation of $TaC_{0.92}$ + χ-$TaRe_{\sim 3.5}$	
	He	2730 ± 25	Melting 80 mol.% $TaC_{0.70}$ + 20 mol.% Re composition results in the formation of $TaC_{0.74}$ + χ-$TaRe_{\sim 1.9}$ *See also* section C–Re–Ta in Table I-2.14	
α-$Ta_{2+x}C$–Re	He	2775 ± 25	Melting 80 mol.% $Ta_{2.06}C$ + 20 mol.% Re composition results in the formation of ~($Ta_{0.55}Re_{0.45}$) solid solution + $TaC_{0.70}$ + χ-$TaRe_{3\pm x}$ (small amount) *See also* section C–Re–Ta in Table I-2.14	[283, 290]
TaC_{1-x}–Rh	–	≥ 1500	The interaction results in the formation of $Rh_{3\pm x}Ta$ and C (graphite) *See also* section C–Rh–Ta in Table I-2.14	[226]
α-$Ta_{2+x}C$–Rh			*See* section C–Rh–Ta in Table I-2.14	

(continued)

Table 2.21 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	References
TaC_{1-x}–Ru	–	~1600	The interaction of non-stoichiometric carbides results in the formation of $TaRu_3C_{1-x}$ ($x \approx 0.6$) *See also* section C–Ru–Ta in Table I-2.14	[226]
α-$Ta_{2+x}C$–Ru			*See* section C–Ru–Ta in Table I-2.14	
TaC_{1-x}–S	–	–	Formation of Ta_2S_2C (2 polytypes; C in carbosulfides is related to graphene sheets on (111) surfaces of TaC_{1-x}) *See also* section C–S–Ta in Table I-2.14	[278]
TaC_{1-x}–Sb			*See* Table 2.24	
TaC_{1-x}–Si	–	–	Formation of SiC and Ta silicides (or ternary phases, including $M_{n+1}AX_n$-phases, ?)	[16, 240, 241, 279, 566]
	–	1500–1700	During spark-plasma sintering the interaction leads to the formation of SiC and $TaSi_2$ *See also* Table 2.24 *See also* section C–Si–Ta in Table I-2.14	
α-$Ta_{2+x}C$–Si			*See* section C–Si–Ta in Table I-2.14	
TaC_{1-x}–Sn			*See* Table 2.24	
TaC_{1-x}–Ta	Vacuum	>1200	Formation of α-$Ta_{2+x}C$ in powdered mixtures	[1, 16, 61, 215, 283, 284, 811]
	–	1900	Under pressure 5 MPa the diffusion welded joint between TaC_{1-x} and Ta is produced (5–15 min exposure)	
	Vacuum	>1900	Formation of α-$Ta_{2+x}C$ and Ta–C solid solutions in the powdered mixtures of components	
	Vacuum	<2200	No interaction between powdered TaC_{1-x} and compact dense Ta (exposure 5 h) *See also* section C–Ta in Table I-2.13	
TaC_{1-x}–δ-TaN_{1-x}–Ta	N_2	1250–1450	Formation of $TaC_{1-x}N_y$ monocarbonitride (cubic) phase, α-$Ta_{2+x}C_yN_z$ and γ-$Ta_{2\pm x}N_yC_z$ semicarbonitride (hexagonal) phases and $Ta(C,N)_{1-x}$ metallic solid solution	[16, 623]
β-$Ta_{2\pm x}C$–Ta		2800–2860	Eutectic β-$Ta_{2\pm x}C$–Ta *See also* section C–Ta in Table I-2.13	[2, 16, 17, 37]
TaC_{1-x}–Tc			*See* section C–Ta–Tc in Table I-2.14	
TaC_{1-x}–Th			*See* section C–Ta–Th in Table I-2.14	
α-$Ta_{2+x}C$–Th			*See* section C–Ta–Th in Table I-2.14	
TaC_{1-x}–α/β-Ti	Vacuum, <7 × 10^{-4} Pa	1750–1800	Molten Ti reacts extensively with TaC_{1-x} crucibles resulting in the formation of an alloy, which soaks into carbide materials; a new phase is apparently formed as a distinct layer based on TiC_{1-x} between carbide and metal Ti	[16, 526]
	–	1830	Formation of mixed phases based on TiC_{1-x} and α-$Ta_{2+x}C$ on the interface *See also* section C–Ta–Ti in Table I-2.14	

(continued)

Table 2.21 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	References
α-$Ta_{2+x}C$–α/β-Ti		1820	Solid solubility of Ti in α-$Ta_{2+x}C$ is negligible	[16, 624]
			See also section C–Ta–Ti in Table I-2.14	
TaC_{1-x}–Tl			*See* Table 2.24	
TaC_{1-x}–$\alpha/\beta/\gamma$-U	–	1300–1310	The penetration depth of molten U into TaC_{1-x} is 0.38 mm (40 min exposure) and 1.02 mm (4 h exposure); $UC_{1\pm x}$ forms	[16]
			See also section C–Ta–U in Table I-2.14	
α-$Ta_{2+x}C$–$\alpha/\beta/\gamma$-U		1700	The max. solid solubility of U in α-$Ta_{2+x}C$ is <7 at.%	[16]
			See also section C–Ta–U in Table I-2.14	
TaC_{1-x}–V			*See* section C–Ta–V in Table I-2.14	
α-$Ta_{2+x}C$–V			*See* section C–Ta–V in Table I-2.14	
TaC_{1-x}–W	–	1700	The addition of 10% TaC_{1-x} disperse inclusions to W leads to increasing the lattice parameter of metallic phase	[1, 37, 61, 215, 283, 286, 287]
	–	>1900–2200	The initiation of reaction and formation of mixed carbide phases	
	–	2000	Under pressure 5 MPa the diffusion welded joint between TaC_{1-x} and W is produced (5–15 min exposure)	
	Vacuum	$\geq$2000	The initiation of reaction between powdered TaC_{1-x} and compact dense W (exposure 5 h)	
	–	2000–2500	Weak interaction between powdered TaC_{1-x} and compact dense metallic W	
	Vacuum	2600	No interaction in the contact zone (3 h exposure) between the compact dense materials	
	He	2600	No interaction in the contact zone (1 h exposure) between the compact dense materials	
	–	>2600	Intensive interaction	
			See also section C–Ta–W in Table I-2.14	
β-$Ta_{2\pm x}C$–W		1700	The max. solid solubility of W in α-$Ta_{2+x}C$ is $\leq$ 7 at.%	[16, 188, 226, 251–253]
		2860	Eutectic β-$(Ta,W)_{2\pm x}C$–(Ta,W)	
			See also section C–Ta–W in Table I-2.14	
TaC_{1-x}–α/β-Zr	–	~1800	Eutectic $(Ta,Zr)C_{1-x}$–(Ta,β-Zr)	[188, 226, 255, 256]
			See also section C–Ta–Zr in Table I-2.14	
α-$Ta_{2+x}C$–Zr			*See* section C–Ta–Zr in Table I-2.14	

[a]The parameters of wettability of TaC_{1-x} phases by liquid metals at various temperatures are listed in Table 2.24

Table 2.22 Chemical interaction and/or compatibility of tantalum carbide phases with refractory compounds at elevated, high and ultra-high temperatures (reaction systems are given mainly in alphabetical order)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	References
TaC_{1-x}–Al_4C_3	–	Up to ~2000	No mutual solid solubilities	[273–275,
	–	–	Formation of Ta_2AlC, Ta_3AlC_2, $Ta_5Al_3C_{1-x}$, Ta_3Al_2C (?) and α/β-Ta_4AlC_3 ($M_{n+1}AX_n$-phases)	513, 514, 689–692, 742]
			See also section TaC_{1-x}–Al in Table 2.21	
			See also section C–Al–Ta in Table I-2.14	
TaC_{1-x}–AlN	Vacuum	1900–2300	High corrosion resistance against AlN sublimation process environment	[567]
TaC_{1-x}–Al_2O_3	CO_2, 3.0×10^{-15} Pa	1000	Calculated equilibrium pressure of the interaction between the components	[471, 472, 565, 770]
	–	~1000–2000	Compatible with each other as composite components	
β-$Ta_{2\pm x}C$– Al_2O_3	–	~1000–2000	Compatible with each other as composite components	[565]
TaC_{1-x}–$B_{4\pm x}C$	–	>1500	Formation of $TaB_{2\pm x}$ and C	[37, 266,
	Ar/He	1900–2400	Addition of 0.3–0.4% $B_{4\pm x}C$ intensifies the densification of powdered carbide by hot-pressing	301, 342]
			See also section C–B–Ta in Table I-2.14	
α-$Ta_{2+x}C$–$B_{4\pm x}C$			*See* section C–B–Ta in Table I-2.14	
TaC_{1-x}–α-BN	–	>2400	Formation of $TaB_{2\pm x}$	[16, 340, 341]
			See also section C–B–N–Ta in Table I-2.14	
TaC_{1-x}–$CeN_{1\pm x}$	–	–	Terminal mutual solid solubilities between the components (?)	[16]
TaC_{1-x}–$CeP_{1\pm x}$	–	–	Terminal mutual solid solubilities between the components (?)	[16]
TaC_{1-x}–$CeS_{1\pm x}$	–	–	Terminal mutual solid solubilities between the components (?)	[16]
TaC_{1-x}–$Cr_7C_{3\pm x}$	–	1000–2000	The max. solubility of "imaginary" phase 'CrC_{1-x}' in TaC_{1-x} is ~10 mol.% and that of TaC_{1-x} in $Cr_7C_{3\pm x}$ is ~1 mol.%	[226–229, 456, 681]
	–	~1720	Eutectic TaC_{1-x}–$Cr_7C_{3\pm x}$	
			See also section C–Cr–Ta in Table I-2.14	
TaC_{1-x}–Cr_2O_3	CO_2, 1.8×10^{-6} Pa	1000	Calculated equilibrium pressure of the interaction between the components	[471, 472]
TaC_{1-x}–$DyN_{1\pm x}$	–	–	Monocarbonitride (cubic) continuous solid solution (complete solubility in the system?)	[16]
TaC_{1-x}–$ErN_{1\pm x}$	–	–	Monocarbonitride (cubic) continuous solid solution (complete solubility in the system?)	[16]
TaC_{1-x}–$EuN_{1\pm x}$	–	–	Terminal mutual solid solubilities between the components (?)	[16]

(continued)

Table 2.22 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	References
TaC_{1-x}–EuO	–	–	Terminal mutual solid solubilities between the components (?)	[16]
TaC_{1-x}–Fe_2O_3	CO_2, 1.5 Pa	1000	Calculated equilibrium pressure of the interaction between the components	[471, 472]
TaC_{1-x}–Fe_3C	–	–	Not (or very slightly) soluble in each other in solid state *See also* section C–Fe–Ta in Table I-2.14	[226, 460]
TaC_{1-x}–$GdN_{1\pm x}$	–	–	Terminal mutual solid solubilities between the components (?)	[16]
TaC_{1-x}–HfC_{1-x}	–	From ~890–905 to 3950	Monocarbide continuous solid solution (data on the higher melting points of $(Ta,Hf)C_{1-x}$ in comparison with individual TaC_{1-x} and HfC_{1-x} are not confirmed); the critical point of the miscibility gap is corresp. to ~$Ta_{0.55–0.58}Hf_{0.42–0.45}C_{1-x}$ composition *See also* section C–Hf–Ta in Table I-2.14	[3, 16, 71, 188, 226, 230, 231, 460, 625–627, 629, 650, 669, 682, 687, 696, 732, 805, 811, 826, 890, 895–897]
TaC_{1-x}–HfC_{1-x}–NbC_{1-x}	–	1900–2000	Monocarbide (cubic) continuous solid solution (complete solubility in the system)	[16, 446, 460]
TaC_{1-x}–HfC_{1-x}–$ThC_{1\pm x}$	–	2000–2100	Terminal solid solution based on TaC_{1-x} – HfC_{1-x} monocarbide (cubic) continuous solid solution; the solubilities of $ThC_{1\pm x}$ in TaC_{1-x} and HfC_{1-x} are <5 mol.% (solubilities of TaC_{1-x} and HfC_{1-x} in $ThC_{1\pm x}$ are very low)	[16]
TaC_{1-x}–HfC_{1-x}–TiC_{1-x}	–	1900–2000	Monocarbide (cubic) continuous solid solution (complete solubility in the system)	[446, 460, 806]
TaC_{1-x}–HfC_{1-x}–$UC_{1\pm x}$	–	1900–2050	Monocarbide (cubic) continuous solid solutions with the miscibility gap limited by ~$(Hf_{0.4}U_{0.6})C_{1-x}$ – ~$(Hf_{0.75}U_{0.25})C_{1-x}$ – ~$(Ta_{0.12–0.15}Hf_{0.55}U_{0.30})C_{1-x}$ compositions *See also* section C–Hf–Ta–U in Table I-2.14	[16, 226, 446, 460]
TaC_{1-x}–HfC_{1-x}–VC_{1-x}	–	1900–2050	Monocarbide continuous solid solutions with great miscibility gap limited by ~$(Hf_{0.98}V_{0.02})C_{1-x}$ – ~$(Hf_{0.03}V_{0.97})C_{1-x}$ – ~$(Ta_{0.45–0.50}Hf_{0.20–0.25}V_{0.25–0.30})C_{1-x}$ compositions	[16, 446, 460, 891]
TaC_{1-x}–HfC_{1-x}–δ-$WC_{1\pm x}$	–	1900–2000	Extended solid solution based on TaC_{1-x} – HfC_{1-x} monocarbide (cubic) continuous solid solution; the solubilities of δ-$WC_{1\pm x}$ in TaC_{1-x} and HfC_{1-x} are ~25–30 mol.% and ~40 mol.%, respectively (the solubilities of TaC_{1-x} and HfC_{1-x} in δ-$WC_{1\pm x}$ are very low)	[16, 446]

(continued)

Table 2.22 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	References
TaC_{1-x}–HfC_{1-x}–ZrC_{1-x}	–	1900–2000	Monocarbide continuous solid solution (complete solubility in the system)	[16, 446]
α-$Ta_{2+x}C$–HfC_{1-x}	–	~1000–2000	The max. solubility of Hf in α-$Ta_{2+x}C$ varies from ~1.5 to ~12 at.% *See also* section C–Hf–Ta in Table I-2.14	[188, 226, 230–231]
β-$Ta_{2\pm x}C$–HfC_{1-x}	–	~2000–3200	The max. solubility of Hf in β-$Ta_{2\pm x}C$ varies from ~1.5 to ~18.5 at.% *See also* section C–Hf–Ta in Table I-2.14	[188, 226, 230, 231]
TaC_{1-x}–$HfN_{1\pm x}$	–	–	Monocarbonitride (cubic) continuous solid solution	[16]
TaC_{1-x}–$HoN_{1\pm x}$	–	–	Monocarbonitride (cubic) continuous solid solution (complete solubility in the system?)	[16]
TaC_{1-x}–$LaB_{6\pm x}$	Vacuum	1550	Interaction leads to the formation of Ta borides (50 h exposure)	[16]
TaC_{1-x}–$LaN_{1\pm x}$	–	–	Terminal mutual solid solubilities between the components (?)	[16]
TaC_{1-x}–$LaP_{1\pm x}$	–	–	Terminal mutual solid solubilities between the components (?)	[16]
TaC_{1-x}–LaS_{1-x}	–	–	Terminal mutual solid solubilities between the components (?)	[16]
TaC_{1-x}–$LuN_{1\pm x}$	–	–	Monocarbonitride (cubic) continuous solid solution (complete solubility in the system?)	[16]
TaC_{1-x}–MgO	–	>2000	Very weak interaction leads to the changes in microhardness near the contact zone	[1, 16, 232, 257, 447]
	Vacuum	>2200–2300	Very weak interaction (with the formation of TaC_xO_y oxycarbide phases) *See also* section C–Mg–O–Ta in Table I-2.14	
TaC_{1-x}–Mn_5C_2, Mn_7C_3	–	–	Not (or very slightly) soluble in each other in the solid state	[460]
TaC_{1-x}–α-MoC_{1-x}	–	2000–2600	Monocarbide (cubic) continuous solid solution *See also* section C–Mo–Ta in Table I-2.14	[21, 226, 235, 236]
TaC_{1-x}–β-$Mo_{2\pm x}C$	–	–	The mutual temporary solubilities of the phases strongly depend on temperature *See also* section C–Mo–Ta in Table I-2.14	[35, 123]
α-$Ta_{2+x}C$–β-$Mo_{2\pm x}C$	–	1500–2200	The solubility of β-$Mo_{2\pm x}C$ in α-$Ta_{2+x}C$ increases from 5 to 30 mol.% and that of α-$Ta_{2+x}C$ in β-$Mo_{2\pm x}C$ – from ~30 to ~60 mol.% with temperature growth *See also* section C–Mo–Ta in Table I-2.14	[188, 276]
α-$Ta_{2+x}C$–β-$Mo_{2\pm x}C$–β-$Nb_{2+x}C$	–	1650–2000	The solubility of β-$Mo_{2\pm x}C$ in α-$Ta_{2+x}C$–β-$Nb_{2+x}C$ semicarbide continuous solid solution is low: from 3–4 mol.% in β-$Nb_{2+x}C$ to 10–18 mol.% in α-$Ta_{2+x}C$; the solubilities of α-$Ta_{2+x}C$ and β-$Nb_{2+x}C$ in β-$Mo_{2\pm x}C$ are 37–47 mol.% and 26–32 mol.%, respectively *See also* section C–Mo–Nb–Ta in Table I-2.14	[188, 276]

(continued)

Table 2.22 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	References
α-$Ta_{2+x}C$–β-$Mo_{2\pm x}C$–β-$V_{2\pm x}C$	–	1650–2000	Semicarbide continuous solid solutions with the miscibility gap limited by ~$(Ta_{0.83–0.90}Mo_{0.10–0.17})_{2\pm x}C$–~$(Ta_{0.37–0.46}Mo_{0.54–0.63})_{2\pm x}C$ – ~$(Ta_{0.30–0.37}Mo_{0.35–0.38}V_{0.25–0.35})_{2\pm x}C$ compositions *See also* section C–Mo–Ta–V in Table I-2.14	[188, 276]
β-$Ta_{2\pm x}C$–β-$Mo_{2\pm x}C$	–	2300–2500	Semicarbide (hexagonal) continuous solid solution *See also* section C–Mo–Ta in Table I-2.14	[21, 226, 235]
TaC_{1-x}–$MoSi_2$	Ar	1850–1950	Interaction in the powder mixtures leads to the formation of $(Mo,Ta)_5Si_{3\pm x}$, SiC and ternary phases *See also* section C–Mo–Si–Ta in Table I-2.14	[16, 35, 123, 233, 234, 733, 894]
TaC_{1-x}–NbC_{1-x}	–	Up to 3600	Monocarbide continuous solid solution (complete solubility in the system, the variation of lattice parameter a, nm with compostion for $(Ta_{1-y}Nb_y)C_{1-x}$ phases ($x \approx 0, 0 \leq y \leq 1$) is linear: $a = 0.4457 + 0.0013y$) *See also* section C–Nb–Ta in Table I-2.14	[16, 188, 226, 231, 238, 399, 460, 626, 627, 685, 741, 811]
TaC_{1-x}–NbC_{1-x}–$ThC_{1\pm x}$	–	2000	Terminal solid solution based on TaC_{1-x}–NbC_{1-x} monocarbide (cubic) continuous solid solution; the solubilities of $ThC_{1\pm x}$ in TaC_{1-x} and NbC_{1-x} are low	[16, 446]
TaC_{1-x}–NbC_{1-x}–TiC_{1-x}	–	1900–2000	Monocarbide (cubic) continuous solid solution (complete solubility in the system) *See also* section C–Nb–Ta–Ti in Table I-2.14	[16, 446, 460, 767]
TaC_{1-x}–δ-TaN_{1-x}–NbC_{1-x}–TiC_{1-x}			Monocarbonitride (cubic) continuous solid solution $(Ta,Nb,Ti)(C,N)_{1-x}$	[767]
TaC_{1-x}–NbC_{1-x}–$UC_{1\pm x}$	–	1900–2000	Monocarbide (cubic) continuous solid solution (complete solubility in the system)	[16, 446, 460]
TaC_{1-x}–NbC_{1-x}–VC_{1-x}	–	1900–2000	Monocarbide (cubic) continuous solid solution (complete solubility in the system)	[16, 446, 460]
TaC_{1-x}–NbC_{1-x}–δ-$WC_{1\pm x}$	–	1450	The max. solubility of δ-$WC_{1\pm x}$ in TaC_{1-x}–NbC_{1-x} continuous solid solution is ~9 mol.%	[16, 280, 446, 685]
	–	1900–2000	The max. solubility of δ-$WC_{1\pm x}$ in TaC_{1-x}–NbC_{1-x} continuous solid solution is ~30 mol.% *See also* section C–Nb–Ta–W in Table I-2.14	
TaC_{1-x}–NbC_{1-x}–ZrC_{1-x}	–	1900–2000	Monocarbide (cubic) continuous solid solution (complete solubility in the system)	[16, 446, 460]
α-$Ta_{2+x}C$–β-$Nb_{2+x}C$	–	–	Semicarbide continuous solid solution *See also* section C–Nb–Ta in Table I-2.14	[188, 226, 231, 238]

(continued)

Table 2.22 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	References
α-$Ta_{2+x}C$–β-$Nb_{2+x}C$–γ-$W_{2\pm x}C$	–	1650–2000	The solubility of γ-$W_{2\pm x}C$ in α-$Ta_{2+x}C$–β-$Nb_{2+x}C$ semicarbide continuous solid solution is low: from 3–4 mol.% in β-$Nb_{2+x}C$ to 11–14 mol.% in α-$Ta_{2+x}C$; the solubilities of α-$Ta_{2+x}C$ and β-$Nb_{2+x}C$ in γ-$W_{2\pm x}C$ are 14–22 mol.% and 24–29 mol.%, respectively *See also* section C–Nb–Ta–W in Table I-2.14	[188, 276]
TaC_{1-x}–δ-NbN_{1-x}	–	–	Monocarbonitride (cubic) continuous solid solution (complete solubility in the system)	[16, 655, 882]
TaC_{1-x}–α-Nb_5Si_{3+x}	–	–	Interaction between the phases leads to the formation of new solid phases, including ternary phases	[16, 35, 123]
TaC_{1-x}–NbO	–	–	Terminal mutual solid solubilities between the components (?)	[16]
TaC_{1-x}–$NdN_{1\pm x}$	–	–	Terminal mutual solid solubilities between the components (?)	[16]
TaC_{1-x}–$Nd_2Fe_{14}B$	Pure Ar	1000	No solubility of Ta in complex boride phase	[799]
TaC_{1-x}–NpC_{1-x}	–	–	Terminal mutual solid solubilities between the components (?)	[16]
TaC_{1-x}–$NpN_{1\pm x}$	–	–	Monocarbonitride (cubic) continuous solid solution (complete solubility in the system?)	[16]
TaC_{1-x}–$PrN_{1\pm x}$	–	–	Terminal mutual solid solubilities between the components (?)	[16]
TaC_{1-x}–PuC_{1-x}	–	1400–1500	Max. terminal solid solubility of TaC_{1-x} in PuC_{1-x} is 17 mol.% and that of PuC_{1-x} in TaC_{1-x} is 67 mol.%	[1, 11, 16, 620, 621]
	–	>1600	Monocarbide continuous solid solution (complete solubility in the system?) *See also* section C–Pu–Ta in Table I-2.14	
TaC_{1-x}–$PuN_{1\pm x}$	–	–	Terminal mutual solid solubilities between the components (?)	[16]
TaC_{1-x}–$PuP_{1\pm x}$	–	–	Terminal mutual solid solubilities between the components (?)	[16]
TaC_{1-x}–$PuS_{1\pm x}$	–	–	Terminal mutual solid solubilities between the components (?)	[16]
TaC_{1-x}–ScC_{1-x}	–	–	Monocarbide continuous solid solution (complete solubility in the system?)	[16, 459]
TaC_{1-x}–$ScN_{1\pm x}$	–	–	Monocarbonitride (cubic) continuous solid solution (complete solubility in the system?)	[16]
TaC_{1-x}–α/β-SiC	–	–	Compatible with each other within the temperature range of the thermal stability of the phases *See also* section C–Si–Ta in Table I-2.14	[1, 16, 245, 725, 893]

(continued)

Table 2.22 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	References
α-$Ta_{2+x}C$–α/β-SiC			*See* section C–Si–Ta in Table I-2.14	
TaC_{1-x}–α/β-Si_3N_4	–	~1400–1600	Interaction leads to the formation of Ta silicide phases and SiC; monocarbide TaC_{1-x} phase inhibits $\alpha \rightarrow \beta$ Si_3N_4 transformation	[795, 796, 821]
TaC_{1-x}–$SmN_{1\pm x}$	–	–	Terminal mutual solid solubilities between the components?	[16]
TaC_{1-x}–$TaB_{2\pm x}$	–	≤ 2100	The components are compatible and virtually insoluble in each other	[14, 16, 24, 266–269]
	–	~2690–2770	Eutectic TaC_{1-x}–$TaB_{2\pm x}$; the max. solid solubility of $TaB_{2\pm x}$ in TaC_{1-x} is ~7 mol.%	
			See also section C–B–Ta in Table I-2.14	
α-$Ta_{2+x}C$–$TaB_{2\pm x}$			*See* section C–B–Ta in Table I-2.14	
TaC_{1-x}–δ-TaN_{1-x}	–	~1500–1900	Monocarbonitride (cubic) continuous solid solution	[15, 63, 226, 237, 449, 570, 645, 889]
			See also section C–N–Ta in Table I-2.14	
TaC_{1-x}–δ-TaN_{1-x}–TiC_{1-x}–δ-$TiN_{1\pm x}$	Pure N_2	1800	Monocarbonitride (cubic) continuous solid solution	[644, 797, 886, 887]
TaC_{1-x}–δ-TaN_{1-x}–'TaO'	–	–	Formation of $TaC_{1-x}N_yO_z$ oxycarbonitride (cubic) phase[a] based on TaC_{1-x}–δ-TaN_{1-x} continuous solid solution	[36]
α-$Ta_{2+x}C$–γ-$Ta_{2\pm x}N$	–	1100–1900	Semicarbonitride continuous solid solution	[226, 449]
			See also section C–N–Ta in Table I-2.14	
α-$Ta_{2+x}C$–γ-$Ta_{2\pm x}N$–'Ta_2O'	–	–	Formation of $Ta_2C_xN_yO_z$ oxycarbonitride (hexagonal) solid solution based on α-$Ta_{2+x}C$–γ-$Ta_{2\pm x}N$ continuous solid solution	[36]
	Ar	1700	Formation of oxycarbonitride (hexagonal) phase $Ta_2C_{0.30–0.44}N_{0.05}O_{0.06–0.23}$ ($0.50 \leq x + y + z \leq 0.60$)	
TaC_{1-x}–α-Ta_5Si_3	–	–	Formation of new solid phases, including ternary phases, in the contact zone	[16, 35, 123]
			See also section C–Si–Ta in Table I-2.14	
TaC_{1-x}–$TaSi_2$	–	–	Formation of new solid phases, including ternary phases; the solubility of TaC_{1-x} in $TaSi_2$ is negligible	[16, 35, 123, 282]
			See also section C–Si–Ta in Table I-2.14	
α-$Ta_{2+x}C$–$TaSi_2$			*See* section C–Si–Ta in Table I-2.14	

(continued)

Table 2.22 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	References
TaC_{1-x}–Ta_2O_5	Vacuum	>1200	Interaction results in the formation of metallic Ta	[36, 215, 449, 530,
	He, Ar 0.1–0.13 MPa	1300–1700	Formation of $TaC_{1-x}O_y$ (TaC_{1-x}–'TaO' oxycarbide (cubic) solid solution) with approximate homogeneity limits: $TaC_{0.7}$–$TaC_{0.7}O_{<0.05}$–$TaC_{0.95}O_{<0.05}$–$TaC_{0.99}$ *See also* section C–O–Ta in Table I-2.14	752, 817]
α-$Ta_{2+x}C$–Ta_2O_5	He, Ar 0.1–0.13 MPa	1300–1700	Formation of $Ta_2C_xO_y$ oxycarbide (hexagonal) solid solution based on α-$Ta_{2+x}C$ with approximate homogeneity limits: $Ta_{2.5}C$–$Ta_2(C_{0.6}O_{0.2})$–$Ta_2(C_{0.4}O_{0.3})$–$Ta_2(C_{0.5}O_{0.4})$–$Ta_{2.0}C$	[36, 449, 752]
	He	1700	Formation of oxycarbide (hexagonal) phase $Ta_2C_{0.25–0.44}O_{0.09–0.24}$ ($0.43 \leq x+y \leq 0.58$) *See also* section C–O–Ta in Table I-2.14	
$TaC_{1-x}TbN_{1\pm x}$	–	–	Monocarbonitride (cubic) continuous solid solution (complete solubility in the system?)	[16]
TaC_{1-x}–TcC_{1-x}	–	–	Monocarbide continuous solid solution (complete solubility in the system?) *See also* section C–Ta–Tc in Table I-2.14	[453, 731]
α-$Ta_{2+x}C$–TcC_{1-x}	–	–	Mutual solid solubilities are low (?) *See also* section C–Ta–Tc in Table I-2.14	[453, 731]
TaC_{1-x}–$ThC_{1\pm x}$ (α-Th, γ-ThC_{2-x})	–	1500	Compatible with each other (mutual solid solubilities are very low) *See also* section C–Ta–Th in Table I-2.14	[16, 226, 448]
TaC_{1-x}–$ThC_{1\pm x}$–TiC_{1-x}	–	>2000	Practically, no mutual solubilities between TaC_{1-x}–TiC_{1-x} monocarbide continuous solid solution and $ThC_{1\pm x}$	[16]
TaC_{1-x}–$ThC_{1\pm x}$–$UC_{1\pm x}$	–	>2000	Monocarbide continuous solid solutions with great miscibility gap because of low mutual solubilities in the TaC_{1-x}–$ThC_{1\pm x}$ system	[16]
TaC_{1-x}–$ThC_{1\pm x}$–VC_{1-x}	–	>2000	Practically, no mutual solubilities between TaC_{1-x}–VC_{1-x} monocarbide continuous solid solution and $ThC_{1\pm x}$	[16]
TaC_{1-x}–$ThC_{1\pm x}$–ZrC_{1-x}	–	>2000	Practically, no mutual solubilities between TaC_{1-x}–ZrC_{1-x} monocarbide (cubic) continuous solid solution and $ThC_{1\pm x}$	[16]
α-$Ta_{2+x}C$–$ThC_{1\pm x}$ (α-Th, γ-ThC_{2-x})	–	1500	Compatible with each other (mutual solid solubilities are very low) *See also* section C–Ta–Th in Table I-2.14	[16, 226, 448]
TaC_{1-x}–$ThN_{1\pm x}$	–	–	Extended monocarbonitride (cubic) solid solutions (terminal solubility?)	[16]
TaC_{1-x}–$ThP_{1\pm x}$	–	–	Terminal mutual solid solubilities between the components (?)	[16]
TaC_{1-x}–$ThS_{1\pm x}$	–	–	Terminal mutual solid solubilities between the components (?)	[16]

(continued)

Table 2.22 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	References
TaC_{1-x}–TiC_{1-x}	–	~0–3070 (or 1600–3070, ?)	Monocarbide continuous solid solution (the variation of lattice parameter a, nm with compostion for $(Ta_{1-y}Ti_y)C_{1-x}$ ($x \approx 0, 0 \leq y \leq 1$) phases is linear: $a = 0.4457 - 0.0128y$)	[16, 188, 226, 237, 246–248, 399, 473, 667, 685,
	–	1700–2000	No complete homogenization is achieved (1 h exposure) for two-phase carbides mixture prepared by carbothermal reduction of oxides	768, 787–789, 811]
			See also section C–Ta–Ti in Table I-2.14	
α-$Ta_{2+x}C$–TiC_{1-x}	–	1500–1800	The max. solubility of Ti in α-$Ta_{2+x}C$ varies from ~6 to ~28 at.%	[246–248]
	–	2000	The max. solubility of Ti in α-$Ta_{2+x}C$ varies from ~2 to ~30 at.%	
			See also section C–Ta–Ti in Table I-2.14	
β-$Ta_{2\pm x}C$–TiC_{1-x}	–	2400–2600	The max. solubility of Ti in β-$Ta_{2\pm x}C$ varies up to ~30 at.%	[246–248]
	–	3000	The max. solubility of Ti in β-$Ta_{2\pm x}C$ varies up to ~20 at.%	
	–	3200	The max. solubility of Ti in β-$Ta_{2\pm x}C$ varies up to ~10 at.%	
			See also section C–Ta–Ti in Table I-2.14	
TaC_{1-x}–TiC_{1-x}–δ-$TiN_{1\pm x}$	Vacuum or Ar	1400–1800	Monocarbonitride (cubic) solid solution $(Ta,Ti)(C,N)_{1-x}$	[644, 773, 779, 787, 788, 886–888]
TaC_{1-x}–TiC_{1-x}–$UC_{1\pm x}$	–	2000–2050	Monocarbide continuous solid solutions with great miscibility gap because of low mutual solubilities in the TiC_{1-x}–$UC_{1\pm x}$ system	[226, 446]
			See also section C–Ta–Ti–U in Table I-2.14	
TaC_{1-x}–TiC_{1-x}–VC_{1-x}	–	1900–2000	Monocarbide (cubic) continuous solid solution (complete solubility in the system)	[446, 460]
TaC_{1-x}–TiC_{1-x}–δ-$WC_{1\pm x}$	–	1400–1500	Extended solid solution based on TaC_{1-x}–TiC_{1-x} monocarbide (cubic) continuous solid solution; the solubilities of δ-$WC_{1\pm x}$ in TaC_{1-x} and TiC_{1-x} are ~10–15 and ~35–50 mol.%, respectively (the solubilities of TaC_{1-x} and TiC_{1-x} in δ-$WC_{1\pm x}$ are very low)	[16, 67, 188, 280, 281, 446, 460, 774, 851, 867, 868]
	–	1900–2500	Extended solid solution based on TaC_{1-x}–TiC_{1-x} monocarbide continuous solid solution; the solubilities of δ-$WC_{1\pm x}$ in TaC_{1-x} and TiC_{1-x} are ~30 and ~65 mol.%, respectively (the solubilities of TaC_{1-x} and TiC_{1-x} in δ-$WC_{1\pm x}$ are very low)	
			See also section C–Ta–Ti–W in Table I-2.14	
TaC_{1-x}–TiC_{1-x}–ZrC_{1-x}	–	1900–2000	Monocarbide (cubic) continuous solid solution (complete solubility in the system)	[16, 446, 460]

(continued)

Table 2.22 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	References
TaC_{1-x}–δ-$TiN_{1\pm x}$	–	–	The mutual solubilities of the phases are limited	[16, 67]
TaC_{1-x}–$TiO_{1\pm x}$	–	–	Terminal mutual solid solubilities between the components (?)	[16]
TaC_{1-x}–$UC_{1\pm x}$	–	1200–1700	Monocarbide continuous solid solution (a spacing-compostion curve in the system is almost linear, complete solubility in the system) *See also* section C–Ta–U in Table I-2.14	[16, 226, 283, 460, 656, 734]
TaC_{1-x}–$UC_{1\pm x}$–VC_{1-x}	–	2000–2050	Monocarbide continuous solid solution with great miscibility gap because of low mutual solubilities in the $UC_{1\pm x}$–VC_{1-x} system *See also* section C–Ta–U–V in Table I-2.14	[16, 226, 446]
TaC_{1-x}–$UC_{1\pm x}$–ZrC_{1-x}	–	1900–2050	Monocarbide (cubic) continuous solid solution (complete solubility in the system) *See also* section C–Ta–U–Zr in Table I-2.14	[16, 226, 446, 460, 898]
α-$Ta_{2+x}C$–$UC_{1\pm x}$			*See* section C–Ta–U in Table I-2.14	
TaC_{1-x}–α-UC_{2-x}	–	1700	The components are compatible; no solubility of TaC_{1-x} in α-UC_{2-x} *See also* section C–Ta–U in Table I-2.14	[283, 291]
TaC_{1-x}–UN_{1-x}	–	–	Monocarbonitride (cubic) continuous solid solution (complete solubility in the system?) *See also* section C–N–Ta–U in Table I-2.14	[16]
TaC_{1-x}–$UP_{1\pm x}$	–	–	Terminal mutual solid solubilities between the components (?)	[16]
TaC_{1-x}–$US_{1\pm x}$	–	–	Terminal mutual solid solubilities between the components (?)	[16]
TaC_{1-x}–VC_{1-x}	–	~500	Practically, no mutual solid solubilities between the components	[3, 16, 226, 227, 231, 399, 460, 627, 682, 736, 811]
	–	From 1200–1300 (or 1700–2500, ?) up to 2700	Monocarbide (cubic) continuous solid solution (complete solubility in the system, the variation of lattice parameter a, nm with compostion for $(Ta_{1-y}V_y)C_{1-x}$ phases ($x \approx 0$, $0 \leq y \leq 1$) is non-linear: $a = 0.4457 - 0.0162y - 0.0123y^2$) *See also* section C–Ta–V in Table I-2.14	
TaC_{1-x}–VC_{1-x}–δ-$WC_{1\pm x}$	–	1900–2000	Extended solid solution based on TaC_{1-x}–VC_{1-x} monocarbide continuous solid solution; the solubilities of δ-$WC_{1\pm x}$ in TaC_{1-x} and VC_{1-x} are ~30 and ~60 mol.%, respectively (the solubilities of TaC_{1-x} and VC_{1-x} in δ-$WC_{1\pm x}$ are very low) *See also* section C–Ta–V–W in Table I-2.14	[16, 446]
TaC_{1-x}–VC_{1-x}–ZrC_{1-x}	–	1900–2000	Monocarbide continuous solid solutions with great miscibility gap limited by ~$(V_{0.99}Zr_{0.01})C_{1-x}$ – ~$(V_{0.03}Zr_{0.97})C_{1-x}$ – ~$(Ta_{0.65}V_{0.20}Zr_{0.15})C_{1-x}$ compositions	[16, 446, 460]

(continued)

Table 2.22 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	References
TaC_{1-x}–δ-VN_{1-x}	–	–	Monocarbonitride (cubic) continuous solid solution (complete solubility in the system?)	[16, 67, 882]
α-$Ta_{2+x}C$–β-$V_{2\pm x}C$	–	–	Semicarbide continuous solid solution	[35, 123, 188, 226]
α-$Ta_{2+x}C$–β-$V_{2\pm x}C$–α-$W_{2+x}C$	–	1650–2000	Semicarbide continuous solid solutions with the great miscibility gap limited by ~$(Ta_{0.86–0.90}W_{0.10–0.14})_{2+x}C$–~$(Ta_{0.21–0.22}W_{0.78–0.79})_{2+x}C$–~$(Ta_{0.18–0.22}W_{0.30–0.35}V_{0.44–0.46})_{2\pm x}C$ compositions *See also* section C–Ta–V–W in Table I-2.14	[188, 276]
TaC_{1-x}–β-W_2B_{5-x}	–	2290	Eutectic TaC_{1-x}–β-W_2B_{5-x}	[580]
TaC_{1-x}–γ-WC_{1-x}	–	~2500–2850	Monocarbide (cubic) continuous solid solution (complete solubility in the system) *See also* section C–Ta–W in Table I-2.14	[16, 188, 226, 236, 249–253]
TaC_{1-x}–δ-$WC_{1\pm x}$	–	1500	The max. solubility of W in TaC_{1-x} corresponds to ~$(Ta_{0.9}W_{0.1})C_{1-x}$ composition	[226, 249, 252, 253, 664, 829]
	–	1800	The max. solubility of W in TaC_{1-x} corresponds to ~$(Ta_{0.8}W_{0.2})C_{1-x}$ composition	
	–	2000	The max. solubility of W in TaC_{1-x} corresponds to compos. ~$(Ta_{0.7}W_{0.3})C_{1-x}$	
	–	2760	The max. solubility of Ta in δ-$WC_{1\pm x}$ is ~4 at.% *See also* section C–Ta–W in Table I-2.14	
TaC_{1-x}–δ-$WC_{1\pm x}$–ZrC_{1-x}	–	1900–2000	Extended solid solution based on TaC_{1-x}–ZrC_{1-x} monocarbide (cubic) continuous solid solution; the solubilities of δ-$WC_{1\pm x}$ in TaC_{1-x} and ZrC_{1-x} are ~30 mol.% (the solubilities of TaC_{1-x} and ZrC_{1-x} in δ-$WC_{1\pm x}$ are very low)	[16, 446]
α-$Ta_{2+x}C$–δ-$WC_{1\pm x}$			*See* section C–Ta–W in Table I-2.14	
TaC_{1-x}–α-$W_{2+x}C$	–	~1250–2100	The mutual solubilities of the phases strongly depend on temperature *See also* section C–Ta–W in Table I-2.14	[16]
α-$Ta_{2+x}C$–α-$W_{2+x}C$	–	1500–1950	The solubility of α-$W_{2+x}C$ in α-$Ta_{2+x}C$ increases from 5 to 12 mol.% and that of α-$Ta_{2+x}C$ in α-$W_{2+x}C$ – from ~8 to ~20 mol.% with temperature growth *See also* section C–Ta–W in Table I-2.14	[188]
β-$Ta_{2\pm x}C$–β-$W_{2+x}C$	–	2300–2450	The solubility of β-$W_{2+x}C$ in α-$Ta_{2+x}C$ increases from ~20 to ~40 mol.% and that of α-$Ta_{2+x}C$ in β-$W_{2+x}C$ – from ~30 to ~50 mol.% with temperature growth *See also* section C–Ta–W in Table I-2.14	[188]

(continued)

Table 2.22 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	References
β-$Ta_{2\pm x}C$–γ-$W_{2\pm x}C$	–	~2460–2790	Semicarbide (hexagonal) continuous solid solution	[16, 188, 226, 251–253]
			See also section C–Ta–W in Table I-2.14	
TaC_{1-x}–$YC_{1\pm x}$	–	–	Terminal mutual solid solubilities between the components (?)	[16]
TaC_{1-x}–$YN_{1\pm x}$	–	–	Terminal mutual solid solubilities between the components (?)	[16]
TaC_{1-x}–$YbN_{1\pm x}$	–	–	Monocarbonitride (cubic) continuous solid solution (complete solubility in the system?)	[16]
TaC_{1-x}–ZrC_{1-x}	–	~940–3400 (or ~1250–3400?)	Monocarbide (cubic) continuous solid solution (complete solubility in the system, the variation of lattice parameter a, nm with compostion for $(Ta_{1-y}Zr_y)C_{1-x}$ phases ($x \approx 0$, $0 \leq y \leq 1$) is non-linear: $a = 0.4457 + 0.0144y + 0.0091y^2$); the critical point of the miscibility gap is corresponding to ~$Zr_{0.30–0.35}Ta_{0.65–0.70}C_{1-x}$ composition	[16, 71, 188, 226, 231, 254–256, 399, 474, 622, 625, 627, 629, 659, 682, 696, 809–814]
	–	<1900	No solid solution is formed (1 h exposure) for two carbides powdered mixture treated by hot pressing	
	Ar, >0.1 MPa	1900–2000	Additions of TaC_{1-x} have no contribution to ZrC_{1-x} hot-pressing densification process, as the formation of solid solutions requires higher temperatures and longer time	
			See also section C–Ta–Zr in Table I-2.14	
α-$Ta_{2+x}C$–ZrC_{1-x}	–	1775–1820	The max. solid solubility of Zr in α-$Ta_{2+x}C$ varies up to ~12 at.%	[16, 188, 226, 231, 255, 256]
			See also section C–Ta–Zr in Table I-2.14	
β-$Ta_{2\pm x}C$–ZrC_{1-x}	–	2500	The max. solid solubility of Zr in β-$Ta_{2\pm x}C$ varies up to ~14 at.%	[16, 188, 226, 231, 255, 256]
			See also section C–Ta–Zr in Table I-2.14	
TaC_{1-x}–$ZrN_{1\pm x}$	–	–	Monocarbonitride (cubic) continuous solid solution (complete solubility in the system)	[16, 67]
TaC_{1-x}–$ZrP_{1\pm x}$	–	–	Terminal mutual solid solubilities between the components (?)	[16]
TaC_{1-x}–β/γ-ZrO_{2-x}	–	≥ 2200	Formation of new solid phases (column-like crystals) in the interface between the compact dense materials	[1, 232, 257, 447, 695]
			See also section C–O–Ta–Zr in Table I-2.14	

[a]Practically, due to the serious difficulties in manufacturing transition metal carbide materials non-contaminated in any degree by O and N, all the materials labeled in literature as TaC_{1-x} with the appropriate certification more likely would have to be considered as $TaC_{1-x}N_yO_z$ with low or very low values of y and z indexes

Table 2.23 Chemical interaction of tantalum carbide phases with gaseous media at elevated, high and ultra-high temperatures (reaction systems are given in alphabetical order)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	References
TaC_{1-x}–CO	CO	–	Formation of oxycarbide phases $TaC_{1-x}O_y$ (extended substitution solid solution based on TaC_{1-x} with approximate homogeneity limits: $TaC_{0.7}$–$TaC_{0.7}O_{<0.05}$–$TaC_{0.95}O_{<0.05}$–$TaC_{0.99}$) *See also* section C–O–Ta in Table I-2.14	[16, 752]
TaC_{1-x}–CO_2	CO_2	–	Formation of oxycarbide phases $TaC_{1-x}O_y$ (extended substitution solid solution based on TaC_{1-x} with approximate homogeneity limits: $TaC_{0.7}$–$TaC_{0.7}O_{<0.05}$–$TaC_{0.95}O_{<0.05}$–$TaC_{0.99}$) and subsequent formation of oxide scales *See also* section C–O–Ta in Table I-2.14	[16, 752]
TaC_{1-x}–Cl_2	Cl_2	<250	Compact dense TaC_{1-x} is resistant to Cl_2	[16, 37, 345]
	Cl_2	700–800	Powdered TaC_{1-x} decomposes easily	
TaC_{1-x}–H_2	H_2, 0.1 MPa	730–1330	The mass loss of compact TaC_{1-x} (6% porosity) observed in the range of 0.02–0.13% (1 h exposure) is a consequence of the formation and subsequent decomposition of the oxide phases (mainly Ta_2O_5) due to residual O_2; in comparison with vacuum (<1.5 Pa) H_2 suppresses the mass loss of TaC_{1-x} (no hydride formation is observed)	[16, 43, 60, 61, 139, 215, 364, 470, 639–642]
	H_2	2000–2500	Weak interaction with the minimally determined mass changes	
	H_2, CH_4	2300–2770	Decarbonization of carbide phases at CH_4 content <0.125–0.250%	
	H_2	~3000–3900	Noticeable decarbonization of carbide phases	
	H_2, CH_4	3180	Decarbonization of carbide phases at CH_4 content <0.25–0.50%	
TaC_{1-x}–H_2O	H_2O[a]	~700	Formation of various complex oxycarbide (?) phases	[139, 449]
	H_2O	800	TaC_{1-x} powders are burning (exothermic reaction)	
TaC_{1-x}–NH_3 (25% N_2 + 75% H_2)	N_2, H_2, CH_4	2300–2500	Decarbonization of carbide phases, if CH_4 content <0.125–0.250%	[215]
	N_2, H_2, CH_4	2770	Decarbonization of carbide phases, if CH_4 content <0.25–0.50%	
	N_2, H_2, CH_4	3180	Decarbonization of carbide phases, if CH_4 content <0.5%	

(continued)

Table 2.23 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	References
TaC_{1-x}–N_2	N_2, 0.1–30 MPa	1100–1800	High level stability; no interaction	[16, 35, 60, 63, 123, 139, 215, 226, 237, 455, 623, 645]
	N_2	1250–1900 (?)	Formation of various monocarbonitride TaC_xN_z phases	
	N_2, $\leq$0.1 MPa	$\leq$2700	No noticeable interaction of TaC_{1-x} materials with N_2 atmosphere	
	N_2	3300	No interaction (if H_2 is absent)	
	N_2	~3900	Decarbonization of carbide phases Data available in literature are controversial *See also* section C–N–Ta in Table I-2.14	
α-$Ta_{2+x}C$–N_2	N_2	1250–1900	Formation of various semicarbonitride $Ta_2C_xN_z$ phases *See also* section C–N–Ta in Table I-2.14	[63, 226, 237, 623]
TaC_{1-x}–O_2[b,c]	O_2 (~0.8 Pa), Ar	50–360	In this temperature range the oxidation of sintered $TaC_{\sim 1.0}$ in gas mixture stream (5 ppm O_2) with heating rate ~1.4 K min^{-1} leads to the maximal reaction rate ~7×10^{-10} mol cm^{-2} s^{-1}; after a prolonged exposition at maximal temperature (8 ppm O_2) the ratio of O/C = ~1.5×10^{-3} was achieved in solids	[1, 16, 36, 61, 139, 215, 240–242, 294–300, 454, 511, 568, 571, 643–645, 652, 703, 752, 771, 817]
	Air	400–500	The oxidation of hot-pressed (1–2% porosity) TaC_{1-x} initiates by the formation of oxycarbide phases $TaC_{1-x}O_y$ (extended substitutional solid solution based on TaC_{1-x} with approximate homogeneity limits: $TaC_{0.70}$–$TaC_{0.70}O_{<0.05}$–$TaC_{0.95}O_{<0.05}$–$TaC_{0.99}$) with the subsequent occurence of pentoxide β-$Ta_2O_5C_x$ (with dissolved carbon in it, ?) and separation of carbon (sometimes metallic Ta was determined in the scale, ?); the oxidation resistance of TaC_{1-x} is higher than that of pure metal Ta	
	Air	~500–700	CVD-prepared TaC_{1-x} coating starts to be oxidized at > 508 °C; the initial oxidation product is $Ta_2O_{3\pm x}$ (hexagonal) solid solution (?)	
	Air	600–700	The oxidation of hot-pressed (1–2% porosity) TaC_{1-x} evolves by the formation of oxide scales containing β-$Ta_2O_5C_x$, suboxide phases TaO_z (sometimes metallic Ta was determined as well, ?) and dispersed carbon phase	
	Air	690–900	During the oxidation of CVD-prepared TaC_{1-x} coating with increasing oxidation temperature $Ta_2O_{3\pm x}$ (hexagonal) solid solution transfoms into β-Ta_2O_5 (orthorhombic) (?)	

(continued)

Table 2.23 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	References
	Air	700–1000	The oxidation rate of hot-pressed TaC_{1-x} grows sharply and oxidation kinetics is described by linear law during long-term exposures; however, oxide scales have good adherence with substrate	
	O_2	700–1000	The non-isothermal oxidation of HIPed $TaC_{0.99}$ (<2% porosity) in dynamic flow of pure O_2 (5.6 $cm^3\ s^{-1}$) at 0.1 MPa at heating rate 1.7×10^{-2} °C s^{-1} leads to increase in oxidation grade (fractional mass change corresponding to the complete oxidation) from 0 to 1	
	O_2	750–850	The isothermal and isobaric (0.02–0.1 MPa) oxidation of HIPed $TaC_{0.99}$ (<2% porosity) in dynamic flow of pure O_2 (5.6 $cm^3\ s^{-1}$); temp. influence is described by apparent activation energy $E = 385 \pm 10$ kJ mol^{-1}	
	Air	800	The oxidation mass gain of TaC_{1-x} is 4.93 and 12.9 g m^{-2} for 1 h and 2 h exposure, respectively	
	Air	900	The oxidation mass gain of TaC_{1-x} is 100 394 g m^{-2} for 1 h and 2 h exposure, respectively	
	Air	~900–1500	The oxidation product on the surface of CVD-prepared TaC_{1-x} coating is β-Ta_2O_5 (orthorhombic) with chapped or porous structure; isolation and/or protective films are not formed	
	Air	1000–1200	The intensive oxidation of hot-pressed TaC_{1-x} accompanied with cracking and splitting of oxide scales; porous samples are disintegrated completely	
	Air	1100–1400	Severe oxidation of materials	
	Air	≥ ~1500	Liquid Ta_2O_5 and β-Ta_2O_5 (orthorhombic) phases coexist on the oxidized surface of CVD-prepared TaC_{1-x} coating	
	O_2, 10^{-1}–10^{-3} Pa	1730–2230	During the oxidation of hot-pressed TaC_{1-x} a metallic surface layer formed by CO release simultaneously reacts with O_2 forming volatile metal oxides; steady-state layer thickness is reached, if the rates of layer growth and metal evaporation are equal in the ablation process	
	O_2/C_2H_2 torch	≥ ~2300	Massive molten Ta_2O_5 (ablation layer; in solidified form – α-Ta_2O_5 + β-Ta_2O_5) wets substrate perfectly (?); the oxidation reaction of TaC_{1-x} CVD-prepared coatings (or hot-pressed materials with porosity <7%) is controlled by the mechanism of oxygen dissolution and diffusion from the gas-liquid interface through the melt to the TaC_{1-x} surface. The liquid oxide layer TaO_z acts as a dynamic component that promotes erosion of the TaC_{1-x} surface (?) *See also* section C–O–Ta in Table I-2.14	

(continued)

Table 2.23 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	References
α-$Ta_{2+x}C$–O_2	Air or O_2	–	Oxidation initiates by the formation of oxycarbide phases $Ta_2C_xO_y$ (extended solid solution based on α-$Ta_{2+x}C$ with approximate homogeneity limits: $Ta_{2.5}C$–$Ta_2(C_{0.6}O_{0.2})$–$Ta_2(C_{0.4}O_{0.3})$–$Ta_2(C_{0.5}O_{0.4})$–$Ta_{\sim 2.0}C$)	[36, 295]
	O_2	700–1000	The non-isothermal oxidation of HIPed α-$Ta_{2+x}C$ (<2% porosity) in dynamic flow of pure O_2 (5.6 $cm^3\ s^{-1}$) at 0.1 MPa at heating rate 1.7×10^{-2} °C s^{-1} leads to increase in oxidation grade (fractional mass change corresponding to the complete oxidation) from 0 to 0.41	
	O_2, 0.02–0.1 MPa	750–850	The isothermal-isobaric oxidation of HIPed α-$Ta_{2+x}C$ (<2% porosity) in dynamic flow of pure O_2 (5.6 $cm^3\ s^{-1}$); temperature influence is described by apparent activation energy $E = 129 \pm 7$ kJ mol^{-1} *See also* section C–O–Ta in Table I-2.14	

[a]In the presence of H_2 in great excess
[b]Fine tantalum carbide powders are pyrophoric [320]
[c]For near-stoichiometric TaC_{1-x} the value of the Pilling-Bedworth ratio $\alpha = M_O d_C / M_C d_O = 1.91$, where M_O is molecular mass of the oxide phase formed on the oxidation of 1 mol of carbide phase, M_C is molecular mass of carbide phase, d_C and d_O are the densities of carbide and oxide phases, respectively [529]

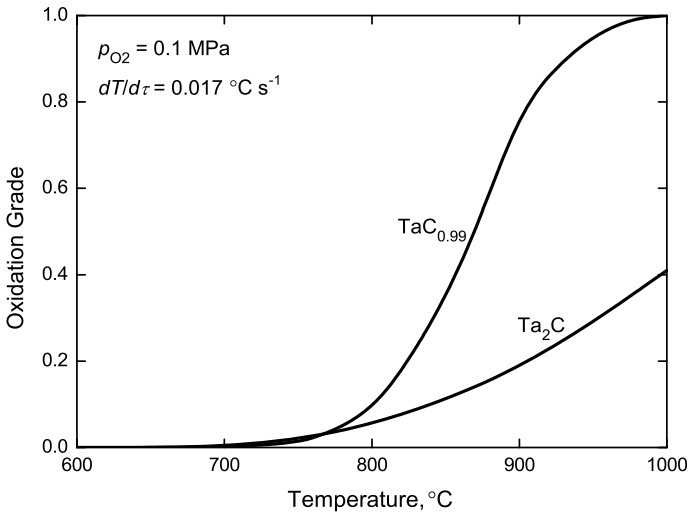

Fig. 2.29 Non-isothermal oxidation kinetics curves (heating rate – 1.7×10^{-2} °C s^{-1}) for 4 × 4 × 4 mm samples of hot isostatically pressed tantalum carbide single-phase materials (porosity <2%) exposed to dynamic flow of pure oxygen (5.6 $cm^3\ s^{-1}$) under pressure 0.1 MPa; oxidation grade was defined by the fractional mass change, which was specified to the mass gain corresponding to the complete oxidation evaluated by considering that tantalum carbides are transformed to Ta_2O_5 according to the following reactions: $2TaC + 4½O_2 \rightarrow Ta_2O_5 + 2CO_2$ or $Ta_2C + 3½O_2 \rightarrow Ta_2O_5 + CO_2$ [295]

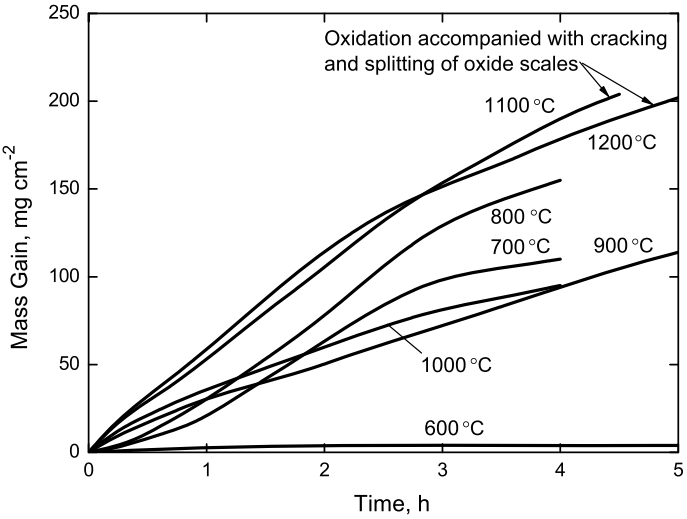

Fig. 2.30 Isothermal oxidation kinetics curves for 8 mm diameter disk-like samples of hot-pressed and subsequently annealed near-stoichiometric tantalum monocarbide materials (porosity 1–2%) in air [242, 296, 297]

Table 2.24 The parameters of wettability of tantalum monocarbide TaC_{1-x} phases with some liquid metals and alloys (melts)[a]

Melt (purity)	Atmo-sphere	Temp., °C	Time, s	$\gamma_{l\text{-}g}$, mJ m^{-2}	W_a, mJ m^{-2}	W_m^b, kJ mol^{-1}	θ, degree	References
				$TaC_{0.78}$				
Cu	Vacuum	1100	1800	1360	1125	–	~100	[37, 258, 259, 347]
				$TaC_{0.90}$				
Cu (0.2% Ag)	Vacuum	1100	900	1270	2325	–	34	[1, 260]
Cu	Vacuum	1100	1800	1360	1595	–	~80	[37, 258, 259, 347]
Cu (0.2% Ag)	Vacuum	1150	900	1255	2355	–	29	[1, 260]
				$TaC_{0.94}$				
Cu (0.2% Ag)	Vacuum	1100	900	1270	1600	–	75	[1, 260]
Cu (0.2% Ag)	Vacuum	1150	900	1255	1845	–	62	[1, 260]
				$TaC_{\sim 1.0}$				
Al	Vacuum	900	900	914	165	8.0	145 ± 2	[261]
Al	Vacuum	900–1000	–	–	–	–	145–146	[259, 346]
Al (99.97%)	Vacuum	950–1000	1200	914	165	–	145	[1]
Al	He (high purity)	1000	900	914	350	–	128 ± 3	[1, 225]
Al (99.97%)	Vacuum	1150	60	914	215	–	140	[1]
Al (99.97%)	Vacuum	1150	180	914	235	–	138	[1]
Al (99.97%)	Vacuum	1150	300	914	325	–	130	[1]
Al (99.97%)	Vacuum	1150	600	914	375	–	126	[1]
Al (99.97%)	Vacuum	1150	1200	914	390	–	125	[1]

(continued)

Table 2.24 (continued)

Melt (purity)	Atmosphere	Temp., °C	Time, s	$\gamma_{l\text{-}g}$, mJ m^{-2}	W_a, mJ m^{-2}	W_m^b, kJ mol^{-1}	θ, degree	References
Al (99.97%)	Vacuum	1200	60	914	455	–	120	[1]
Al (99.97%)	Vacuum	1200	120	914	1225	–	70	[1]
Al (99.97%)	Vacuum	1200	180	914	>1825	–	~0	[1]
Al	He (high purity)	1200	900	865	395	–	123 ± 3	[1, 225]
Al	He (high purity)	1400	900	865	940	–	85 ± 3	[1, 225]
Bi (99.999%)	Vacuum	320	900	390	114	5.4	135.0 ± 0.3	[1, 261, 346]
Co[c]	Vacuum	1420	300	1910	3770	–	13 ± 2	[260, 262, 346]
Co (99.98%)	Vacuum, Ar	1500	900	1805	>3670	–	~0	[1, 259, 261, 346]
Co[d]	Vacuum	1500	1200	1850	3700	–	10	[262, 263]
Cu	Vacuum	1100					~60	[37, 258, 347]
Cu	Vacuum	1100	–	–	–	–	75	[16, 362, 646]
Cu (0.2% Ag)	Vacuum	1100	900	1270	1525	–	78	[1, 260, 875]
Cu	Vacuum	1130	900	1351	318	15.1	140 ± 1	[1, 261, 346]
Cu (0.2% Ag)	Vacuum	1150	900	1255	1685	–	69	[1, 260]
Cu	Vacuum	1150	–	–	–	–	78	[346]
Cu (0.2% Ag)	Vacuum	1200	900	1240	1860	–	60	[1, 260]
Cu	Vacuum	1250	–	–	–	–	60	[346]
Cu	Vacuum	1250	–	–	–	–	36	[16, 362, 646]
Fe	Vacuum	1450–1490	60	1900	3650	–	23	[1, 259, 260, 346]
Fe (C-4.0, Mn-1.4, Si-2.5, S-0.1%)[e]	Pure Ar	1400–1450	900	–	–	–	~0	[1, 304, 346]
Fe (C-3.1, Mn-0.4, Si-1.6, S-0.02%)[e]	Pure Ar	1400–1450	900	–	–	–	~0	[1, 304, 346]
Fe (C-2.7, Mn-0.4, Si-1.2, S-0.2%)[e]	Pure Ar	1400–1450	900	–	–	–	~0	[1, 304, 346]
Fe (C-2.6, Mn-1.0, Si-1.0, Cr-0.3%)[e]	Pure Ar	1400–1450	900	–	–	–	~0	[1, 304, 346]
Fe (C-1.0, Cr-1.4, Mn-0.4, Si-0.3%)[e]	Pure Ar	~1500	900	–	–	–	~0	[304, 346]

(continued)

Table 2.24 (continued)

Melt (purity)	Atmo-sphere	Temp., °C	Time, s	$\gamma_{l\text{-}g}$, mJ m^{-2}	W_a, mJ m^{-2}	W_m^b, kJ mol^{-1}	θ, degree	References
Fe (C-1.0, Cr-1.4, Mn-1.1, Si-0.6%)[e]	Pure Ar	~1500	900	–	–	–	~0	[304, 346]
Fe (C-0.8, Mn-0.2, Si-0.2, S-0.03%)[e]	Pure Ar	1500–1550	900	–	–	–	~0	[1, 304, 346]
Fe (C-0.2, Cr-14.5, Ni-2.7, Mn-0.5%)[e]	Pure Ar	1500–1550	900	–	–	–	~0	[1, 304, 346]
Fe (99.999%)	Vacuum, Ar	1550	900	1780	>3560	–	~0	[1, 261, 346]
Ga (99.99999%)	Vacuum	800	900	707	250	12.1	130 ± 2	[1, 261, 346]
Ge (99.9999%)	Vacuum	1000	900	600	230	11.0	128 ± 1	[1, 261, 346]
In (99.9995%)	Vacuum	250	900	559	56	2.7	154 ± 3	[1, 261, 346]
Mn (99.8%)	Vacuum, Ar	1300	900	1750	>3500	–	~0	[1, 261]
Ni	Vacuum	1380	300	1810	3550	–	16	[1, 260, 346]
Ni (99.99%)	Vacuum, Ar	1400	900	1700	>3400	–	~0	[1, 259, 261, 346]
Pb (99.98%)	Vacuum	400	900	480	171	8.2	130 ± 1	[1, 261, 346]
Sb (99.999%)	Vacuum	700	900	384	198	9.6	119 ± 3	[1, 261, 346]
Si (99.9999%)	Vacuum	1500	900	860	>1720	–	~0	[1, 261, 346]
Sn (99.999%)	Vacuum	300	900	554	130	6.2	140 ± 1	[1, 261, 346]
Tl (99.999%)	Vacuum	400	900	490	126	6.1	138 ± 1	[1, 261, 346]

[a]The parameters of wettability are given in accordance with Young-Dupré equation $W_a = \gamma_{l\text{-}g} \times (1 + \cos\theta)$ and Young's equation $\gamma_{s\text{-}l} = \gamma_{s\text{-}g} - \gamma_{l\text{-}g}\cos\theta$, where W_a is the work of adhesion, $\gamma_{l\text{-}g}$ is the liquid-vapour interfacial energy (surface tension), $\gamma_{s\text{-}l}$ is solid-liquid interfacial energy, $\gamma_{s\text{-}g}$ is the solid-vapour interfacial energy and θ is the wetting contact angle [1]; compositions of melts are given in mass (weight) percentage

[b]$W_m = W_a(M/d)^{2/3}N_A^{1/3}$, where W_m is the molar work of adhesion, M is the molecular mass and d is the density of chemical compound, N_A is the Avogadro constant [264]

[c]$\gamma_{s\text{-}l} = 550 \pm 200$ mJ m^{-2}

[d]$\gamma_{s\text{-}l} = 0.905$ mJ m^{-2}

[e]Sintered materials (porosity – 12-25%)

Table 2.25 Diffusion rates and related parameters in the systems containing tantalum, carbon and tantalum carbides phases at various temperatures[a]

Species pair	Temperature dependence of the diffusion coefficient (diffusivity) $D = D_0\exp[(-E_A/R)/T]$, $cm^2\ s^{-1}$	Temperature range, K (°C)	Remarks on materials characteristics and measurement method	References
C → Ta	$6.7 \times 10^{-3}\exp(-19{,}400/T)$	460–2950 (190–2680)	Combined data from several sources	[35, 123, 307, 492]
	$3.8 \times 10^{-2}\exp(-24{,}300/T)$	2470–2950 (2200–2680)	Radiotracer ^{14}C with mechanical sectioning	[35, 123, 307]
	$2.57 \times 10^{-2}\exp(-21{,}500/T)$	1470–1870 (1200–1600)	^{14}C radiometric method, sectioning technique	[16, 501]
	$1.2 \times 10^{-2}\exp(-20{,}300/T)$	1720–2470 (1450–2200)	Radiotracer ^{14}C with mechanical sectioning	[502]
	$2.78 \times 10^{-3}\exp(-12{,}400/T)$	870–1670 (600–1400)	Radiotracer ^{14}C with mechanical sectioning	[503]
	$4 \times 10^{-3}\exp[(-16{,}500 \pm 90)/T]$	415–535 (140–260)	From the measurements of internal friction	[16, 308]
	$1.5 \times 10^{-2}\exp[-(12{,}600–13{,}600)/T]$	320–430 (50–160)	From various measurements of internal friction	[495–497]
	$(6.1 \pm 1.2) \times 10^{-3}\exp[(-19{,}400 \pm 130)/T]$	460–630 (190–360)	From the measurements of internal friction	[35, 123, 308, 499, 505]
	$[(0.54–1.40) \pm (0.1–0.3)] \times 10^{-2} \times \times\exp\{[-(19{,}300–19{,}900) \pm (100–130)]/T\}$	460–630 (190–360)	Determined by the various methods of internal friction measurements	[308, 495, 498, 499]
	$8.54 \times 10^{-4}\exp(-26{,}400/T)$	1870–2170 (1600–1900)	–	[37]
	$(4.2^{+6.1}_{-2.5})\exp[(-46{,}000 \pm 2{,}000)/T]$	1970–2470 (1700–2200)	Carbidization of metal Ta; calculated from the diffusion coefficients obtained from layer growth	[318]
	$84.6\exp(-49{,}700/T)$	1570–1870 (1300–1600)	Calculation based on the growth of TaC_{1-x} phase layer	[504]

(continued)

Table 2.25 (continued)

Species pair	Temperature dependence of the diffusion coefficient (diffusivity) $D = D_0\exp[(-E_A/R)/T]$, $cm^2\ s^{-1}$	Temperature range, K (°C)	Remarks on materials characteristics and measurement method	References
	$1.98 \times 10^4\exp[(-9{,}700 \pm 3{,}300)/T]$, ?	1270–2070 (1000–1800)	Calculation based on the growth of α-$Ta_{2+x}C$ phase layer	[493–495]
	$3.43 \times 10^{-6}\exp(-10{,}500/T)$	1270–2070 (1000–1800)	Calculation based on the thickness of α-$Ta_{2+x}C$ layer	[16, 90]
	$9 \times 10^{-3}\exp(-21{,}100/T)$	–	Theoretical values evaluated by using the elasic constants of tantalum and diameter of carbon atom	[495, 500]
	$\sim\exp(-15{,}100/T)$	900–1000 (630–730)	Studied by high-temperature high-resolution Auger spectroscopy during the carbonization of Ta ribbons	[679]
C → TaC_{1-x}	$3.9\exp[(-59{,}700 \pm 5{,}900)/T]$	2630–3230 (2360–2960)	Polycrystalline (hot-pressed) TaC_{1-x} ($x \approx 0.02$–0.08, $\sim Ta(C_{0.93}N_{0.04}O_{0.01})_{0.98}$, 15–20 μm grain size, 7–8% porosity, O + N ≤ 0.6%), ^{14}C radiotracer	[41, 123, 305, 314, 316, 508]
	1.66×10^{-10} 2.58×10^{-9} 1.66×10^{-8}	2500 (2230) 2800 (2530) 3100 (2830)	Calculated on the basis of developed statistic and thermodynamic theory of interstitial phases (for $x = 0.02$)	[559]
	$1.38\exp(-45{,}100/T)$	2370–2920 (2100–2650)	Parameters of temperature variation of TaC_{1-x} layer growth rate constant upon contact saturation of solid metallic Ta by C	[35, 123, 531]
	$2.00\exp(-45{,}600/T)$	2370–2920 (2100–2650)	Parameters of reaction-chemical diffusion, contact saturation of solid Ta by C; layer growth metallography method	[1, 37, 123, 309, 509]
	$1.04\exp(-43{,}300/T)$	2070–2970 (1800–2700)	Parameters of reaction-chemical diffusion, contact saturation of solid Ta by C; layer growth metallography method	[1, 123, 310, 531]
	$0.18\exp(-43{,}000 \pm 1{,}500/T)$	1970–2970 (1700–2700)	Parameters of reaction-chemical diffusion, contact saturation of solid Ta by C in the area of compos. TaC_{1-x} ($0.02 \le x \le 0.04$) (initial composition with $x = 0.34$); X-ray method	[1, 37, 123, 311, 316, 350, 509, 531]
	$0.135(-42{,}300/T)$	1970–2970 (1700–2700)	Parameters of reaction-chemical diffusion; diffusion couple ($TaC_{0.84}$-C) method	[349, 350]

(continued)

Table 2.25 (continued)

Species pair	Temperature dependence of the diffusion coefficient (diffusivity) $D = D_0\exp[(-E_A/R)/T]$, $cm^2\ s^{-1}$	Temperature range, K (°C)	Remarks on materials characteristics and measurement method	References
	$4.88 \times 10^{-4}\exp(-28{,}400/T)$	2560–2810 (2290–2540)	Parameters of reaction-chemical diffusion, rate of carbidization of Ta obtained by deposition from gas phase on surface of graphite ($TaC_{0.89}$); X-ray method	[123, 312]
	$8.8\exp(-49{,}700/T)$	2470–3020 (2200–2750)	Parameters of reaction-chemical diffusion, contact saturation of solid Ta by C; layer growth metallography method	[123, 313]
	$8.9 \times 10^{-2}\exp(-32{,}400/T)$	2100–2900 (1830–2630)	Chemical diffusion coefficients calculated on the basis of experiments on deposition of non-stoichiometric carbide TaC_{1-x} ($0.10 \leq x \leq 0.34$, porosity $\leq 10\%$) from $CH_4 + H_2$ gas phase[b]	[1, 37, 123, 315]
	$2.67\exp(-50{,}900/T)$	2370–2970 (2100–2700)	Carbidization of 99.97% purified metal Ta in Ar atmosphere (equivalent single phase diffusion); gravimetric techniques	[316, 317]
	$4.9 \times 10^{-5}\exp(-17{,}100/T)$	2370–2970 (2100–2700)	Carbidization of 99.97% purified metal Ta in Ar – 30 vol.% N_2 atmosphere (equivalent single phase diffusion); gravimetric techniques	[316, 317]
	4.75×10^{-8}	2870 (2600)	Carbidization of 99.97% purified metal Ta in Ar atmosphere (two phase diffusion); gravimetric techniques	[316, 317]
	$(2.3^{+1.6}_{-0.9}) \times 10^{-2}\exp[(-42{,}000 \pm 1{,}000)/T]$	1970–2470 (1700–2200)	Carbidization of metal Ta; calculated from the diffusion coefficients obtained by fitting a modified error function on the measured concentration profiles (for $TaC_{\sim 1.0}$)	[318]
	$(1.1^{+2.8}_{-0.8})\exp[(-49{,}000 \pm 2{,}000)/T]$	1970–2470 (1700–2200)	Carbidization of metal Ta; calculated from the diffusion coefficients obtained from layer growth	[318]
	$(0.14^{+0.25}_{-0.09})\exp[(-42{,}000 \pm 2{,}000)/T]$	1970–2470 (1700–2200)	Carbidization of metal Ta; calculated from the diffusion coefficients obtained by fitting a modified error function on the measured concentration profiles (for $TaC_{0.77}$)	[318]
	$\sim\exp[(-49{,}100 \pm 600)/T]$	2000–2500 (1730–2230)	Parameters of reaction-chemical diffusion measured by diffusion couples method	[649]
	$4 \times 10^{-3}\exp(-38{,}100/T)$	–	The values of chemical diffusion coefficients summarized on the basis of several works	[372]

(continued)

Table 2.25 (continued)

Species pair	Temperature dependence of the diffusion coefficient (diffusivity) $D = D_0\exp[(-E_A/R)/T]$, $cm^2\ s^{-1}$	Temperature range, K (°C)	Remarks on materials characteristics and measurement method	References
C → ζ-Ta_4C_{3-x}	$(20^{+11}_{-7})\exp[(-58{,}500 \pm 1{,}000)/T]$	1970–2420 (1700–2150)	Carbidization of metal Ta; calculated from the diffusion coefficients obtained from layer growth	[318, 372]
C → β-$Ta_{2\pm x}C$	$10^3\exp(-57{,}900/T)$	2370–2920 (2100–2650)	Parameters of reaction-chemical diffusion, contact saturation of solid Ta by C; metallography method	[1, 37, 123, 309, 316, 509]
	$7.0\exp(-44{,}800/T)$	2070–2970 (1800–2700)	Parameters of reaction-chemical diffusion, contact saturation of solid Ta by C; metallography method	[1, 123, 310]
	$2.7\exp(-42{,}800/T)$	2470–3020 (2200–2750)	Parameters of reaction-chemical diffusion, contact saturation of solid Ta by C; metallography method	[123, 313]
	2.97×10^{-7}	2870 (2600)	Carbidization of 99.97% purified metal Ta in Ar atmosphere (two phase diffusion); gravimetric techniques	[316, 317]
	$(1.1^{+4.2}_{-0.9}) \times 10^5\exp[(-76{,}000 \pm 3{,}000)/T]$	1970–2470 (1700–2200)	Carbidization of metal Ta; calculated from the diffusion coefficients obtained by fitting a modified error function on the measured concentration profiles and from layer growth as well	[318, 372]
	$10^{-5}\exp(-75{,}400/T)$	–	The values of chemical diffusion coefficients summarized on the basis of several works	[372]
Ta → TaC_{1-x}	$24.6\exp(-61{,}200/T)$	–	Values calculated on the diffusion theory basis	[1, 37, 348]
	$\sim\exp(-62{,}400/T)$	–	General evaluation	[169, 170, 521]
	$\sim\exp(-76{,}300/T)$	–	Estimated on the basis of LeClaire's relation (connected with phase melting point)	[531]
W → TaC_{1-x}	$(1.4^{+6.5}_{-1.2})\exp[(-60{,}400 \pm 4{,}000)/T]$	2270–2820 (2000–2550)	Polycrystalline TaC_{1-x} ($x \approx 0.4\%$ porosity, O + N ≤ 0.2%), ^{185}W radiotracer	[41, 306, 507]
W → TaC_{1-x}, W → $(W,Ti,Ta)C_{1-x}$	–	–	The mixed carbide systems have the solid solutions, which possess a maximum value of the activation energy of W diffusion	[825]

(continued)

Table 2.25 (continued)

[a]The chemical diffusion coefficient concentration and temperature dependencies can be represented by an equation of the form $D' = D'_0\exp(\beta x)\exp[(-E'_A/R)/T]$, where β is the concentration factor and x is the value of index in TaC_{1-x} formula (for the ranges of x = 0.01–0.43 and T = 1730–2730 °C the recommended values are $D'_0 = 0.136$ cm^2 s^{-1}, $\beta = 6.0$ and $E'_A = 358$ kJ mol^{-1} [350]), and the discrepancy between chemical and self-diffusivities may be estimated on the basis of the thermodynamical relationship: $D' = D \times [1 + (d\ln\gamma_C/d\ln N_C)]$, where γ_C is the activity coefficient of carbon and N_C is the atomic fraction of carbon (evaluated at 2000 °C in TaC_{1-x}: at x = 0.01, $d\ln\gamma_C/d\ln N_C = 134$ and at x = 0.20, $d\ln\gamma_C/d\ln N_C = 9.4$ [506]); the approximate values of apparent activation energy for some diffusion controlled processes in TaC_{1-x}, α-$Ta_{2+x}C$ and β-$Ta_{2\pm x}C$: (**a**) formation kinetics of α-$Ta_{2+x}C$ during the $2TaC_{\sim 1.0}$ + Ta interaction – 220 ± 20 kJ mol^{-1} (900–1200 °C) [711–713]; (**b**) formation kinetics of ζ-Ta_4C_{3-x} (x = 0.44) during the $2TaC_{\sim 1.0}$ + Ta interaction – 1010 ± 150 kJ mol^{-1} (1600–1800 °C) [711–713]; (**c**) recrystallization (grain growth) – 380 kJ mol^{-1} (sintered $TaC_{\sim 1.0}$, 2400–2700 °C) [177]; (**d**) powder hot-pressing densification – 210 kJ mol^{-1} (β-$Ta_2C_{\sim 1.0}$, viscous flow, 2700–3050 °C) [515, 516], 210 kJ mol^{-1} (β-$Ta_2C_{\sim 1.0}$, mean particle size 15–25 μm, 2700 °C) [647], 295 kJ mol^{-1} ($TaC_{0.73}$, mean particle size 15–25 μm, 2700 °C) [647], 325 kJ mol^{-1} ($TaC_{0.82}$, mean particle size 15–25 μm, 2700 °C) [647], 360 kJ mol^{-1} ($TaC_{0.91}$, mean particle size 15–25 μm, 2700 °C) [647], 405 kJ mol^{-1} ($TaC_{\sim 1.0}$, viscous flow, 2700–3050 °C) [515, 516]; (**e**) indentation – 465 kJ mol^{-1} (single crystal $TaC_{0.83}$, 1200–1300 °C) [169]; (**f**) internal friction peaks characteristics (Marx-Wert formula) – 360 kJ mol^{-1} (sintered $TaC_{0.99}$, 1640 °C) [537]; data on creep – *see* Sect. 2.4 (Table 2.16), *see also* Sect. 2.5 (Table 2.20)

[b]The parameters probably contain significant contribution of surface and grain-boundary diffusion due to rather high porosity of the materials

Table 2.26 The interaction of near-stoichiometric tantalum monocarbide TaC_{1-x} materials with some chemical reagents in aqueous solutions [1, 37, 43, 90, 215, 319, 320, 345, 352, 353, 362, 512, 532]

Reagent, formula (density or concentration of aqueous solution)[a]	Treatment conditions		Character of interaction[b]
	Temperature, °C	Exposure time, h	
HCl (1:1)	20	24	No decomposition
	110–112	2	Decomposes up to ~2%
HCl (d = 1.19)	20	24	No decomposition
	120	2	Decomposes up to ~2%
H_2SO_4 (1:4)	20	24	Decomposes up to ~1%
	115	2	Decomposes up to ~7%
H_2SO_4 (d = 1.84)	20	24	No decomposition
	255	1.5	Decomposes up to ~75% (10% amorphous C) with the formation of stable oxysulfate; composition of released gases: CO_2 96.0, CO 2.2 and H_2 1.6 mol.% (CO_2 and CO are products of the oxidation of amorphous C)[c]
	275	1.0	Decomposes up to ~80% (6.0% amorphous C) with the formation of stable oxysulfate; composition of released gases: CO_2 92.3, CO 6.0 and H_2 1.7 mol.% (CO_2 and CO are products of the oxidation of amorphous C)[c]
	290–310	1.5–2.0	Decomposes completely (solution is transparent finally)[c]
HNO_3 (1:1)	20	24	Decomposes up to ~1%
	105	2	Decomposes up to ~2%
HNO_3 (d = 1.43)	20	24	No decomposition
	110–114	1–2	Decomposes up to ~1%
H_3PO_4 (1:3)	≥100	2	Decomposes up to ~2%
H_3PO_4 (d = 1.70)	20	24	Decomposes up to ~2%
	150–300	2	Decomposes very slightly
$H_2C_2O_4$[d] (saturated solution)	20	24	Decomposes up to ~3%
	100–105	2	Decomposes up to ~2%
NaOH (10%)	20	24	Decomposes up to ~1%
	108	2	No decomposition
NaOH (20%)	20	24	Decomposes up to ~1%
	108	2	Decomposes up to ~1%
H_2O_2 (30%)	110	1	Decomposes up to ~15–17%
NH_4F (5%)	110	1	Decomposes up to ~10%
$(NH_4)_2S_2O_8$ (25–30%)	110	1	Decomposes up to ~10–12%
3HCl (1:1) + HNO_3 (1:2)	110		No decomposition
3HCl (d = 1.19) + HNO_3 (d = 1.43)	20	24	Decomposes up to ~1%
	110–115	2	Decomposes up to ~2%
HCl (d = 1.19) + H_2O_2 (30%)	110	1	Decomposes up to ~1%

(continued)

Table 2.26 (continued)

Reagent, formula (density or concentration of aqueous solution)[a]	Treatment conditions		Character of interaction[b]
	Temperature, °C	Exposure time, h	
HCl (d = 1.19) + $(NH_4)_2S_2O_8$ (25%)	–	–	Decomposes up to ~2%
HCl (d = 1.19) + Br_2 (HBrO, HBr) (saturated solution)	–	–	Decomposes up to ~2%
H_2SO_4 (d = 1.84) + HNO_3 (d = 1.43)	20	24	Decomposes up to ~9%
	110	2	Decomposes up to ~4%
H_2SO_4 (1:4) + H_3PO_4 (1:3)	20	24	No decomposition
	≥150	2	Decomposes up to ~1%
H_2SO_4 (d = 1.84) + H_3PO_4 (d = 1.70)	20	24	Decomposes up to ~2%
	≥150	2	Decomposes ≥50% with the deposition of salts and hydrolysis
H_2SO_4 (1:4) + $H_2C_2O_4$[d] (saturated solution)	20	24	Decomposes up to ~7%
	≥100	2	Decomposes up to ~2%
H_2SO_4 (d = 1.84) + $H_2C_2O_4$[d] (saturated solution)	20	24	Decomposes up to ~3%
	≥100	2	Decomposes up to ~3%
H_2SO_4 (1:4) + H_2O_2 (30%)	110	1	Decomposes up to ~20–25%
H_2SO_4 (1:4) + $(NH_4)_2S_2O_8$ (25%)	110	1	Decomposes up to ~2–5%
(1–4)HNO_3 (d = 1.43) + HF (d = 1.15)[e]	20	–	Decomposes completely
	110	0.5	Decomposes completely
HNO_3 (d = 1.43) + NH_4F (5%)	110–120	0.5	Decomposes completely; composition of released gases: CO_2 95–97, CO (traces) and H_2 2–3 mol.% (no precipitated C, solution is transparent)
H_2O_2 (30%) + $H_2C_4H_4O_6$[f] (50%)	110	1	Decomposes up to ~40–50%
H_2O_2 (30%) + $H_3C_6H_5O_7$[g]	110	1	Decomposes up to ~43%
H_2O_2 (30%) + $H_4C_{10}H_{12}N_2O_8$[h] (saturated solution)	110	1	Decomposes up to ~96%
H_2O_2 (30%) + NH_4F	110	1	Decomposes up to ~96%
$(NH_4)_2S_2O_8$ + NH_4F	–	–	Decomposes up to ~37%
4NaOH (20%) + Br_2 (HBrO, HBr)	20	24	No decomposition
	100–110	2	Decomposes ≥50% with the deposition of salts
4NaOH (20%) + H_2O_2 (30%)	20	24	Decomposes up to ~38%
	≥100	2	Decomposes ≥50% with the deposition of salts and hydrolysis
4NaOH (20%) + $K_3[Fe(CN)_6]$ (10%)[i]	20	24	Partly decomposes by hydrolysis
	≥100	2	Decomposes ≥50%

[a]All the ratios are given in volume fractions

[b]When it is not indicated specially, the character reported is related to the powders with mean particle size of 40–50 μm

[c]Represented by the reaction equation: $2TaC_{1-x} + 4H_2SO_4 + 5SO_3 = Ta_2O(SO_4)_4 + 2(1 - x)C + 5SO_2 + 4H_2O$

[d]Oxalic acid

[e]α-$Ta_{2+x}C$ decomposes completely in this solvent as well [37]

[f]Tartaric acid

(continued)

Table 2.26 (continued)

[g]Citric acid
[h]Ethylenediaminetetraacetic acid (EDTA)
[i]Recommended chemical etching agents for TaC_{1-x}: (**a**) 10 g $K_3[Fe(CN)_6]$ + 2–10 g KOH/NaOH + 100 ml H_2O (Murakami's reagent) [442]; (**b**) boiling aqueous solutions KOH (20%) + $K_4[Fe(CN)_6]$ (20%) with exposure 3 min [347]; (**c**) boiling mixture of concentrated H_2SO_4 + $(NH_4)_2S_2O_8$ [441]; (**d**) H_2NSO_3H (20% sulfamic acid) with exposure 1 min (for thin films on different substrates) [442, 443]; (**e**) HF + 40% NH_4F (BHF solution) with exposure 4 s (for thin films on different substrates) [442, 443]; (**f**) 1 part HF (40%) + 3 parts concentrated HNO_3 with exposure 2 min [510]; (**g**) 40 ml HCl + 10 ml HNO_3 + 1 ml HF at 60–70 °C with exposure 4 min [381]; (**h**) 10 ml HNO_3 + 1 ml HF + 0.5 ml lactic acid $CH_3CH(OH)CO_2H$ at 60–70 °C with exposure 0.5–1.0 min (for determination of the density of dislocations) [159]

oxidation kinetics of near-stoichiometric TaC_{1-x} materials given in the latter one can be termed in the context of ridge-effect model proposed by Shabalin [303, 452] with ridge temperature at air common conditions (for a certain partial oxygen pressure) nearby 800 °C, since the kinetics curves at temperatures ≥1000 °C are not reproducible and representative because of randomly cracking and splitting oxide scales. Data on the catalytic activities of tantalum carbides are given in several works [1, 351, 354, 355, 475–490, 630–639, 764, 800, 827, 899]; the examples of electrochemical behaviour of tantalum carbides are described in [355, 491, 639, 827]. The parameters of wettability of tantalum monocarbide phases with some liquid metals (melts) are listed in Table 2.24, the diffusion rates for the systems containing tantalum carbide phases are presented in species pairs within the various ranges of temperatures in the Table 2.25. The characters of chemical interaction of tantalum monocarbide phases with some common chemicals (acids, alkalies and salts in aqueous solutions) are summarized in Table 2.26.

In comparison with other ultra-high temperature materials the summarized data on the chemical behaviour of tantalum carbides are given in Addendum.

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Chapter 3
Hafnium Monocarbide

3.1 Structures

Practically hafnium forms with carbon the only one chemical compound (*see* also section C–Hf in Table I-2.13) – hafnium monocarbide HfC_{1-x} with extremely broad homogeneity range, apart from low-temperature (<510–530 °C) ordered and metastable structures of $Hf_{2\pm x}C$ (*Fd(–3)m*, *R(–3)m*, *P3m1*, *Pnma*, *I4$_1$/amd*, *Pbcn*, *P4/mmm*), $Hf_3C_{2\pm x}$ (*C2/m*, *I4/mmm*, *P(–3)m1*, *Immm*), $Hf_4C_{3\pm x}$ (*C2/c*, *P(–3)m1*, *Pm(–3)m*), $Hf_5C_{4\pm x}$ (*P(–1)*, *C2/m*, *I4/m*), $Hf_6C_{5\pm x}$ (*C2/m*, *C2/c*, *P3$_1$12*), $Hf_7C_{6\pm x}$ (*R(–3)*) and $Hf_8C_{7\pm x}$ (*P4$_3$32*, *Fm(–3)m*) as well as molecular cluster Hf_8C_{12}, which are not confirmed sufficiently in literature [1–7, 332, 333, 483, 540, 541, 548]. The ordered structures should be in thermodynamic equilibrium with disordered hafnium monocarbide HfC_{1-x} phase; however, the latter one can exist as a metastable phase at a temperature below ~500 °C for an infinitely long time because the diffusion rate in this temperature range is extremely small [2]. A high-temperature partial variant of hafnium – carbon phase diagram is given in Fig. 3.1, and the microstructural features of hafnium monocarbide HfC_{1-x} are presented in Table 3.1. The C/Hf radii ratio calculated on the basis of Pauling's atomic size of Hf (0.1585 nm, CN = 12) is 0.486 [9], 0.46 [32]; the ratio of Hf radii (in nm) in Me/MeC is 0.159/0.164 (2.5% expansion of Hf atoms in carbide) [42]. The phase transformation of HfC_{1-x} from NaCl type (*Fm(–3)m*) to CsCl type structure (*Pm(–3)m*) under very high pressures (~100–500 GPa) has been predicted theoretically [315, 322, 323, 326, 528].

The variation of the lattice parameter of hafnium monocarbide HfC_{1-x} phases with carbon content (or deviation from the stoichiometry) within the homogeneity range is shown in Fig. 3.2; in a modified form the following equations described this relationship in HfC_{1-x} were proposed by Senczyk (for $0 \leq x \leq 0.3$) [581]:

$$a = 0.4637 - 0.0017x - 0.01425x^2, \tag{3.1}$$

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https://doi.org/10.1007/978-94-024-1302-1_3

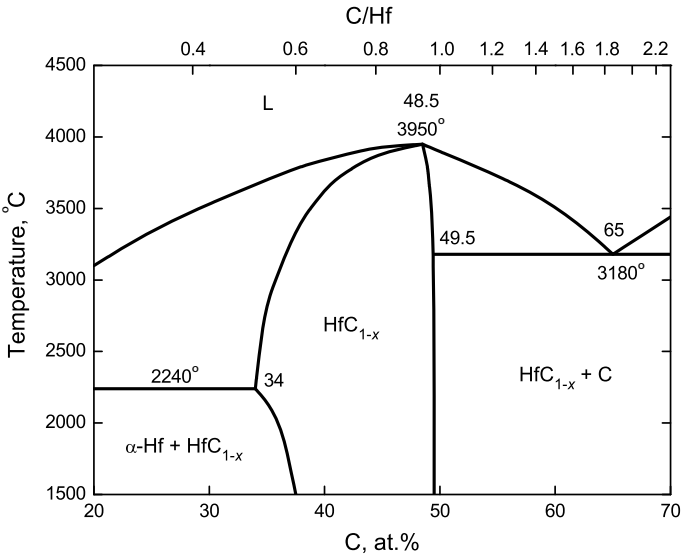

Fig. 3.1 High-temperature partial variant of hafnium – carbon equilibrium phase diagram [2, 4, 6, 8–15, 17, 201, 343, 344, 426, 500, 559, 632]

and Gusev (for $x \leq 0.36$) [587]:

$$a = 0.4639 - 0.00125x - 0.00138x^2, \tag{3.2}$$

where a is lattice parameter, nm and x is the value of index in HfC_{1-x} formula. The wide data spread available in literature for the relationships between the lattice parameter and composition of HfC_{1-x} mainly could be explained by some difficulties in the contamination control of phase composition; the considerable solubilities of oxygen and nitrogen, which are always present in HfC_{1-x}, result in the formation of oxycarbide $HfC_{1-x}O_y$, carbonitride $HfC_{1-x}N_z$ and oxycarbonitride $HfC_{1-x}N_zO_y$ phases (*see also* sections C–Hf–N, C–Hf–N–O and C–Hf–O in Table I-2.14) with various deviations from the stoichiometry and noticeable decrease in the magnitude of lattice parameter [13, 28, 40, 223, 226, 402–408, 619]; the examples of oxygen effect on the completeness of metal and non-metal sublattices in HfC_{1-x} ($HfC_{1-x}O_y$) are given in Table 3.2.

The strong covalent character of chemical bonding in HfC_{1-x} determines the high value of C–Hf bond energy (enthalpy), which is about 5.60 ± 0.3 eV [165, 495, 496]. The calculated parameters of the electron structures of HfC_{1-x} monocarbides with minimal and maximal deviations from the stoichiometry are presented in Table 3.3.

Hafnium monocarbide has a uniquely stable crystal lattice at high temperatures compared with other transition metal carbides. HfC_{1-x} exhibits a marked anisotropy in its behaviour; in the <100> direction, it is relatively plastic compared with its

Table 3.1 Structural properties (crystal structure, density) of hafnium monocarbide phase

Formula	Crystal structure					Density[c], g cm^{-3}	References
	System	Type	Space group	Lattice parameter[a] *a*, nm	Z[b]		
HfC_{1-x}	Cubic	NaCl	*Fm(–3)m*	0.4619[d]	4	–	[12, 33]
				0.4620[e]	4	–	[28]
				0.4623[f]	4	–	[26]
				0.4629[g]	4	–	[38]
				0.4630	4	12.74	[18]
				0.4630[h]	4	12.47	[13]
				0.4631[i]	4	–	[30]
				0.4633	4	12.71	[19]
				0.4635	4	12.70	[27, 32]
				0.4635[j]	4	12.67	[34, 40]
				0.4637[g]	4	–	[39]
				0.4637	4	12.68	[29]
				0.4637[k]	4	12.53	[13]
				0.4638[l]	4	12.60	[16]
				0.4639[m]	4	12.66	[24, 37, 560]
				0.4640	4	12.66	[9, 20, 21, 35, 36]
				0.4640[n]	4	12.65	[3, 4, 8]
				0.4641	4	12.65	[6, 17, 31]
				0.4642[o]	4	12.67	[1, 13–15]
				0.4642[p]	4	12.52	[557]
				0.4643	4	12.63	[22]
				0.4648	4	12.59	[23]
				0.4674[q] (?)	4	–	[25]

[a]When it is not indicated specially, value reported is for near-stoichiometric composition
[b]Number of formula units per lattice cell
[c]Calculated from XRD or neutron diffraction patterns
[d]Carbon content – 37.9 at.% (in equilibrium with α-Hf); the experimentally measured variation of lattice parameter with temp.: 20 °C – 0.4619 nm, 490 °C – 0.4632 nm, 680 °C – 0.4638 nm, 890 °C – 0.4643 nm
[e]Samples also contain some O and N
[f]Samples also contain some Zr
[g]Calculated on the basis of density-functional theory (DFT) with local density approximation (LDA)
[h]Carbon content – 40.0 at.%
[i]Minimal interatomic distances: Hf–Hf and C–C – 0.327446 nm, Hf–C – 0.231539 nm
[j]Carbon content – 48.7 at.%
[k]Carbon content – 45.0 at.%
[l]Carbon content – 47.4 at.%
[m]Reported thermal increment – 4.1×10^{-6} K^{-1}
[n]Carbon content – 49.7 at.% (in equilibrium with graphite)
[o]Carbon content – 49.2 at.%
[p]Prepared by SHS with charged C/Hf ratio – 1.0, carbon content – 47.4 at.%
[q]In the presence of excess carbon at temp. >1700 °C

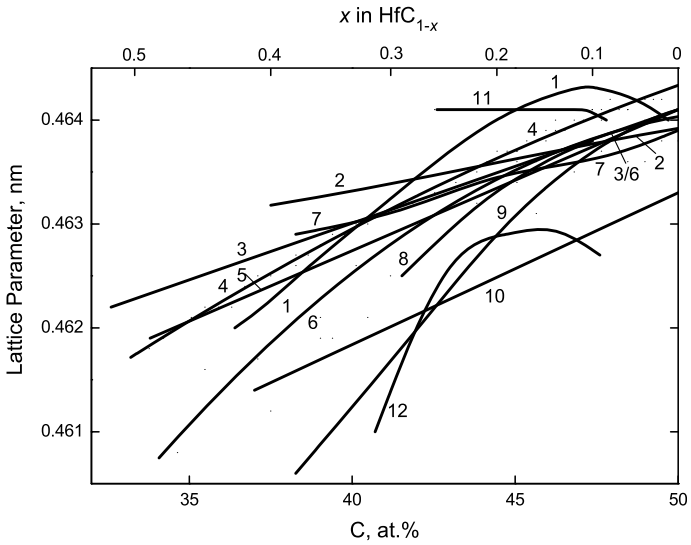

Fig. 3.2 Lattice parameter of HfC_{1-x} as a function of phase composition: 1 – synthesized by various methods [124, 558], 2 – [99], 3 – [17], 4 – polynomial approximation [14, 15], 5 – [22, 40], 6 – [8], 7 – synthesized from the elements [84, 423, 477, 561], 8 – [16], 9 – [6], 10 – [19], 11 – synthesized in loosely sintered hydride-carbon powdered mixtures in high vacuum (content O – ~0.04%) and 12 – the same method realized in the compacted mixtures of the same powders (content O – ~0.07%) [584, 585] (1, 4–6 – on the basis of several sources)

more brittle behaviour in the <110> direction, where cleavage occurs [147]. At room and low temperatures for monocarbide $HfC_{0.98}$ the preferred slip system (110) <1 $\bar{1}$ 0> was determined by the experiments with single crystal materials [41, 42].

For parameters of formation and migration of lattice point defects (vacancies and interstitial atoms) in hafnium monocarbide *see* Sect. 3.5 (Table 3.17). Recently, due to the massive progress in nanotechnology, some nanostructures based on HfC_{1-x},

Table 3.2 The completeness of hafnium monocarbide HfC_{1-x} ($HfC_{1-x}O_y$) sublattices with various oxygen contents [40, 406]

Phase composition	Fraction of structural defects in sublattices	
	Metal	Non-metal
$HfC_{0.76}O_{0.07}$	0.030	0.195
$HfC_{0.76}O_{0.26}$	0.0525	0.0325
$HfC_{0.75}O_{0.24}$	0.055	0.0625
$HfC_{0.74}O_{0.25}$	0.0525	0.0625
$HfC_{0.69}O_{0.30}$	0.0625	0.075
$HfC_{0.66}O_{0.12}$	0.0625	0.2675
$HfC_{0.66}O_{0.21}$	0.0525	0.175
$HfC_{0.57}O_{0.13}$	0.0575	0.340

Table 3.3 The calculated parameters of the electron structures of hafnium monocarbide HfC_{1-x} phases with different deviations from the stoichiometry [490]

Parameter	Compostion	
	$HfC_{\sim 1.0}$	$HfC_{0.67}$
Electron configuration of Hf atom	$6s^{0.31}\ 6p^{0.93}\ 5d^{1.75}$	$6s^{0.46}\ 6p^{0.94}\ 5d^{1.89}$
Electron configuration of C atom	$2s^{1.17}\ 2p^{3.84}$	$2s^{1.25}\ 2p^{3.80}$
Charge of Hf atom	+1.0	+0.71
Charge of C atom	−1.0	−1.06
Populations of (p + d) states of the Hf atom	2.68	2.83
SWASC[a] d^5, %	35.0	38.0
SWASC[a] sp^3, %	37.4	36.9
Dissociation energy between Hf–Hf atoms, eV	0.230	0.695
Breakaway energy of Hf–Hf atoms, eV	7.62	10.25
Interaction energy Hf–C atoms, eV	0.82	0.48

[a]Statistical weighting of the atoms with stable configuration

e.g. single-crystalline nanotubes (~60 nm in diameter and ~15 nm wall thickness) [398], nanowires (20–80 nm in diameter, several tens of micrometers in length) grown along <100> and <112> crystal directions [43, 44, 562, 563, 637], nanocrystal chains [399], nanofibers (~40–90 nm in diameter, up to several micrometers in length) [512], whiskers [375–377], thin films and coatings [148, 534, 593, 597, 598, 602, 606–610, 612, 614–616], nanopowders and nanoparticles [473, 508, 520, 635], have been synthesized and studied; 2D-molecular Hf_2C MXene [519, 533, 537, 555, 556] was examined using first principles calculations.

The recommended value for the bulk density of pure poreless near-stoichiometric HfC_{1-x} materials at room temperature is 12.60–12.65 g cm^{-3} [1, 4, 31, 35, 36, 45, 525].

3.2 Thermal Properties

Hafnium monocarbide HfC_{1-x} has one of the highest melting points of all the solid substances available. Within the homogeneity range of HfC_{1-x} the melting point of the phase varies (*see* Fig. 3.1); the maximum temperature is pertaining to the nonstoichiometric compositions around ~$HfC_{0.89–0.94}$ [8, 21, 46, 47], and there are no sufficient proofs for the complex carbides to be outperforming the maximum melting point of hafnium monocarbide phase [46]. The general thermodynamic properties of near-stoichiometric hafnium monocarbide are summarized in Table 3.4. For the molar heat capacity $c_p = f(T, \text{K})$, J mol^{-1} K^{-1}, of hafnium monocarbide phases with various deviations from the stoichiometry, the following relationships were recommended in literature:

Table 3.4 General thermodynamic properties of near-stoichiometric monocarbide HfC_{1-x}

Characteristics	Symbol	Unit	Value	References
Standard heat of formation (at 298.15 K)[a]	$-\Delta H°_{298}$	kJ mol^{-1}	208.4 ± 7.5	[1, 49]
			209.2	[51]
			209.5 ± 1.5[b]	[8, 9, 56, 412]
			219.0	[383]
			223.0	[48]
			226.8 ± 2.1	[35, 50, 69, 490]
			228.1 ± 4.2	[36]
			230.1	[31, 55]
			234.1 ± 10.0[c]	[2, 577–579]
			251.0	[54]
			309.0	[32, 45]
			338.0	[53]
			339.1	[52]
Standard molar entropy[d] (at 298.15 K and 100 kPa)	$S°_{298}$	J mol^{-1} K^{-1}	28.1 ± 8.2	[36]
			39.46[e]	[8, 9]
			39.64	[58, 59]
			40.08	[1]
			41.22 ± 2.10	[31, 35, 54, 55]
			47.73[f]	[45, 57]
Molar enthalpy difference	$H_{298} - H_0$	kJ mol^{-1}	6.355	[58]
			6.373	[8, 9]
Standard molar heat capacity[g] (at 298.15 K and 100 kPa)	$c°_{p,298}$	J mol^{-1} K^{-1}	33.4	[61]
			34.42 ± 2.10	[29, 31, 54, 55]
			35.30	[45, 60]
			37.49[e]	[4, 8, 9, 58, 59, 62]
			40.38	[1]
			54.7	[36]
Specific heat capacity (at 298.15 K)	c	J kg^{-1} K^{-1}	180.7	[31]
			196.8	[4]
Molar enthalpy (heat) of vaporization (at 298.15 K)[h,i]	ΔH_v	kJ mol^{-1}	1568 ± 2	[74]
Melting point	T_m	K (°C)	4100 (3830)[j]	[14, 15, 22, 382, 383]
			4120 ± 40 (3850 ± 40)	[63]
			4160 ± 150 (3885 ± 150)	[45, 64–67]
			4165 ± 150 (3890 ± 150)	[32, 35, 36, 53, 68, 384–386]

(continued)

Table 3.4 (continued)

Characteristics	Symbol	Unit	Value	References
			4170 (3900)[k]	[135, 420]
			4200 ± 40 (3930 ± 40)[l]	[4, 9, 12, 13, 61, 82, 387, 410, 411]
			4220 ± 25 (3950 ± 25)	[8, 12, 31, 388]
			4230 ± 20 (3960 ± 20)	[1]
Boiling point	T_b	K (°C)	4430 (4160)	[50]
			5190 (4920)	[410]
			5670 (5400)[f]	[32, 45]

[a]Enthalpy (heat) of complete dissociation (atomization) from solid state at 298,15 K ($-\Delta_{at}H°_{298}$, kJ mol^{-1}): 1546 ± 12 [1], 1640 [51], 1627 [409], 1674 [57], 1570 ± 10 [579]
[b]Carbon content – 48.9 at.%
[c]Extrapolated to the stoichiometric composition
[d]Molar entropy $S°_T$ (at 1200 K), J mol^{-1} K^{-1}, for HfC_{1-x} phases with different deviations from the stoichiometry: 103.17 (x = 0), 102.49 (x = 0.05), 101.80 (x = 0.10), 101.11 (x = 0.15), 100.42 (x = 0.20), 99.73 (x = 0.25), 99.05 (x = 0.30), 98.36 (x = 0.35) and 97.67 (x = 0.40) [58]
[e]Carbon content – 49.5 at.%
[f]Calculated
[g]Molar heat capacity $c_{p,T}$ (at 1200 K), J mol^{-1} K^{-1}, for HfC_{1-x} phases with different deviations from the stoichiometry: 49.93 (x = 0), 49.34 (x = 0.05), 48.76 (x = 0.10), 48.17 (x = 0.15), 47.58 (x = 0.20), 47.00 (x = 0.25), 46.41 (x = 0.30), 45.83 (x = 0.35) and 45.24 (x = 0.40) [58]
[h]Due to an insignificant change in the hafnium monocarbide phase composition during congruent vaporization, the calculation was carried out in accordance to the second and third laws of thermodynamics with the assumption of phase composition constancy
[i]Enthalpy (heat) of sublimation of metallic Hf at 298,15 K, $\Delta_{sMe}H°_{298}$ = 622.4 kJ mol^{-1} [580]
[j]Carbon content – 47.5 at.%
[k]Carbon content – 49.7 at.%
[l]Carbon content – ~48.5 at.%

for $HfC_{0.71}$ (in the range of temperatures from 1300 to 2500 K) [58, 69, 70]

$$c_p = 42.74 + \left(5.532 \times 10^{-3}\right)T, \tag{3.3}$$

for $HfC_{0.85}$ (in the range of temperatures from 1300 to 2500 K) [58, 69, 70]

$$c_p = 42.96 + \left(6.076 \times 10^{-3}\right)T, \tag{3.4}$$

for $HfC_{0.98}$ (in the range of temperatures from 298 to 3000 K) [4, 8]

$$c_p = 44.07 + \left(9.18 \times 10^{-3}\right)T - \left(12.748 \times 10^{-7}\right)T^2 - \left(8.181 \times 10^5\right)T^{-2}, \tag{3.5}$$

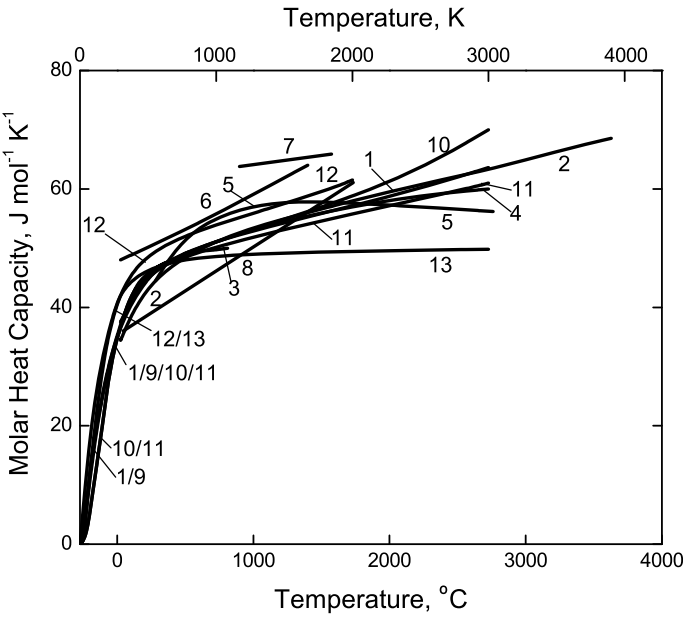

Fig. 3.3 Variation of molar heat capacity (1–11 – at constant pressure, c_p, and 12–13 – at constant volume, c_v) with temperature for near-stoichiometric hafnium carbide HfC_{1-x} (1 – [58, 59]; 2 – [54, 55]; 3 – [29]; 4 – [8]; 5 – [139]; 6 – [539]; 7 – $HfC_{0.96}$, contents: non-combined C – 0.03%, Zr – 3.45% [307]; 8 – [82, 309]; 9 – $HfC_{0.97}$, contents: non-combined C – 0.014%, O – 0.003%, N – 0.031%, Zr – 0.035%, Fe – 0.005% [491]; theoretically calculated: 10 – Slater's approximation and 11 – Dugdale-MacDonald's approximation on the basis of Debye-Grüneisen model [62]; 12 – *ab initio* calculated, based on the mean values between the corresponding generalized gradient approximation (GGA) and local density approximation (LDA) estimations [336]; 13 – using projector augmented-wave (PAW) method within the GGA of the Perdew-Burke-Ernzerhof scheme for ionic solids (PBEsol) [442]; 1–2, 4, 8 – on the basis of several sources; when it is not indicated specially, data given are for quasi-stoichiometric composition)

for $HfC_{0.99}$ (in the range of temperatures from 1300 to 2500 K) [58, 69–71]

$$c_p = 43.19 + \left(6.670 \times 10^{-3}\right)T \tag{3.6}$$

and for HfC_{1-x} in general (in the range of temperatures from 1300 to 2500 K) [58]

$$c_p = 38.957 + \left(9.142 \times 10^{-3}\right)T - 11.716x, \tag{3.7}$$

or

$$c_p = 41.018 + \left(8.174 \times 10^{-3}\right)T + 7.738 \times 10^{-3}xT - 27.134x, \tag{3.8}$$

where x is the value of index in HfC_{1-x} formula. For the specific heat capacity of $HfC_{0.98}$ in the range of temperatures from 1290 to 2800 K the equation $c = f(T, \text{K})$, J kg^{-1} K^{-1} is given by Levinson [72] as

$$c = 243.4 + \left(27.73 \times 10^{-6}\right)T. \tag{3.9}$$

The variations of molar heat capacity c_p with temperature for near-stoichiometric hafnium monocarbide are demonstrated on the basis of several sources in Fig. 3.3. The thermal properties of HfC_{1-x} are very sensitive to the deviation from the stoichiometry: for the standard heat of formation $\Delta H°_{298}$, kJ mol^{-1}, molar enthalpy difference $H°_T - H°_{298}$, J mol^{-1}, and molar entropy $S°_T$, J mol^{-1} K^{-1}, the following concentration and temperature-concentration dependencies within the homogeneity range of HfC_{1-x} were calculated from the experimental data [2, 49, 58, 73, 577–579]:

$$\Delta H°_{298} = -208.4 - 69.36x + 662.85x^2, \tag{3.10}$$

$$\Delta H°_{298} = 42.52 - 378.0(1-x) + 101.4(1-x)^2 \pm 10.0, \tag{3.11}$$

$$H°_T - H°_{298} = 38.957T + \left(4.571 \times 10^{-3}\right)T^2 - 15724x - 11.716xT - 9600 \tag{3.12}$$

$$H°_T - H°_{298} = 41.018T + \left(4.087 \times 10^{-3}\right)T^2 + \left(3.689 \times 10^{-3}\right)xT^2 - 27.134xT - 11744 \tag{3.13}$$

$$S°_T = 89.72\lg T + \left(9.142 \times 10^{-3}\right)T + 69.321x - 26.98x\lg T - 184.06 \tag{3.14}$$

where T is temperature, K, and x is the value of index in HfC_{1-x} formula (Eqs. (3.12)–(3.14) were recommended for the range of temperatures from 1200 to 2500 K). The thermodynamic functions of stoichiometric hafnium carbide are tabulated by Turchanin et al. [58, 59] in the range of 0–3000 K, by Schick [55] in the range of 0–3900 K and by Barin [54] in the range of 298.15–3900 K, the thermodynamic functions of $HfC_{0.98}$ in the range of 298.15–3000 K – by Storms [8] and Toth [9], and $HfC_{0.97}$ in the range of 5–350 K – by Westrum and Feick [491].

During the vaporization processes from the surface of hafnium monocarbide at high and ultra-high temperatures in vacuum, the composition of the carbide phase (C/Hf ratio) can change noticeably. The following equation was recommended for hafnium partial pressure (P_{Hf}, Pa) over near-stoichiometric carbide phase in equilibrium with carbon (HfC_{1-x} + C) by Storms [8] for 2500–3200 K (2200–2900 °C):

$$\lg P_{Hf} = -\left(4.206 \times 10^4\right)/T + 13.306, \tag{3.15}$$

the estimation of hafnium partial pressures (P_{Hf}, Pa) in the gaseous phase in the Hf-C system carried out by Kulikov [410] led to such relationships as

for the conditions of congruent dissociation of quasi-stoichiometric HfC_{1-x}:

at 1000–2500 K (730–2230 °C)

$$\lg P_{\mathrm{Hf}} = -\left(4.013 \times 10^4\right)/T + 12.620, \tag{3.16}$$

at 2500–4200 K (2230–3930 °C)

$$\lg P_{\mathrm{Hf}} = -\left(4.076 \times 10^4\right)/T + 12.878, \tag{3.17}$$

at 4200–5000 K (3930–4730 °C)

$$\lg P_{\mathrm{Hf}} = -\left(3.397 \times 10^4\right)/T + 11.262, \tag{3.18}$$

for the conditions of HfC_{1-x}–C phases equilibrium:

at 1000–2500 K (730–2230 °C)

$$\lg P_{\mathrm{Hf}} = -\left(4.438 \times 10^4\right)/T + 12.675, \tag{3.19}$$

at 2500–4200 K (2230–3930 °C)

$$\lg P_{\mathrm{Hf}} = -\left(4.372 \times 10^4\right)/T + 12.414, \tag{3.20}$$

at 4200–5000 K (3930–4730 °C)

$$\lg P_{\mathrm{Hf}} = -\left(3.683 \times 10^4\right)/T + 10.773, \tag{3.21}$$

where T is temperature, K (*see* Table 3.5). The partial pressures of hafnium and carbon over the carbide phase as functions of composition at the fixed temperature of 3000 K (2730 °C) are shown in Fig. 3.4 in comparison with the equilibrium pressure of carbon over graphite surface. For the evaporation rate of $HfC_{0.95}$ G, g cm^{-2} s^{-1}, as a function of temperature T, K, the following equation was proposed for the range of temperatures from 2600 K (2330 °C) to 3290 K (3020 °C) [1, 69]:

$$\lg G = 8.41 - \left(3.916 \times 10^4\right)/T. \tag{3.22}$$

For monocarbide phase $HfC_{0.99}$ at 2600 and 2900 °C the values of evaporation rate were estimated as 2.7×10^{-6} and 6.4×10^{-5} g cm^{-2} s^{-1} [1, 69], while for carbide $HfC_{0.98}$ Deadmore [83] measured at 2225 and 2625 °C the values of 7×10^{-8} and 6×10^{-6} g cm^{-2} s^{-1}, respectively. For their high-vacuum experiments at 2800 K (2530 °C) Hojo and Nakayama [118] have evaluated the vapour pressure over a emitter coated by near-stoichiometric HfC_{1-x} and its evaporation rate as 8.29×10^{-3} Pa and 7.17×10^{-7} g cm^{-2} s^{-1}, respectively. Through a large part of the homogeneity range of HfC_{1-x} the vapour phase should receive metallic atoms mainly, although in a certain area of compositions congruent vaporization may occur [14, 15]. The change in the C/Me ratio of hafnium monocarbide phase established that at 2900 K (2630 °C) the congruently vaporizing composition corresponds to $HfC_{0.94}$ [20, 75], Nikolskaya and Avarbe [74] showed that this composition varies from $HfC_{0.99}$ at 2350 K (2080 °C) to $HfC_{0.91}$ at 3290 K (3020 °C), and in the temperature range of 2690–3290 K (2420–3020 °C) it could be described by the equation as follows:

Table 3.5 Parameters of the gaseous phase in the Hf–C system in the conditions of Hf–HfC_{1-x} phases equilibrium, congruent dissociation of quasi-stoichiometric HfC_{1-x} and HfC_{1-x}–C phases equilibrium calculated on the basis of thermodynamic data [410]

Parameters	Temperature, K (°C)								
	1000 (730)	1500 (1230)	2000 (1730)	2500 (2230)	3000 (2730)	3500 (3230)	3900 (3630)	4200 (3930)	5000 (4730)
					Hf−HfC_{1-x} phases equilibrium				
lgP_{Hf}, Pa	−20.164	−9.413	−4.064	−0.896	1.111	2.493	3.296	3.930	–
lgP^{a}_{Σ}, Pa	−20.164	−9.413	−4.064	−0.896	1.111	2.495	3.305	3.945	–
lg(C/Hf)[b]	−15.656	−9.821	−6.813	−4.905	−3.470	−2.388	−1.661	−1.453	–
Contents, vol.%:									
C	–	–	–	–	0.02	0.27	1.44	2.10	–
C_2	–	–	–	–	–	–	0.02	0.04	–
C_3	–	–	–	–	–	–	–	0.003	–
Hf	100.0	100.0	100.0	100.0	99.97	99.59	97.87	96.62	–
HfC[c]	–	–	–	–	0.01	0.14	0.67	1.24	–
					Congruent dissociation of quasi-stoichiometric HfC_{1-x}				
lgP_{Hf}, Pa	−27.493	−14.339	−7.520	−3.425	−0.702	1.242	2.439	3.174	4.468
lgP^{a}_{Σ}, Pa	−27.192	−14.038	−7.219	−3.126	−0.407	1.530	2.721	3.454	4.758
Contents, vol.%:									
C	50.00	50.00	49.92	49.32	47.71	43.60	39.46	37.13	36.16
C_2	–	–	0.04	0.31	1.12	2.46	3.64	4.44	5.07
C_3	–	–	0.005	0.07	0.39	1.07	1.71	2.12	1.68
C_4	–	–	–	–	–	–	–	0.005	0.01
C_5	–	–	–	–	–	–	–	–	0.002
Hf	50.00	50.00	50.03	50.23	50.69	51.58	52.30	52.44	51.31
HfC[c]	–	–	0.004	0.07	0.39	1.29	2.59	3.85	5.77

(continued)

Table 3.5 (continued)

Parameters	Temperature, K (°C)								
	1000 (730)	1500 (1230)	2000 (1730)	2500 (2230)	3000 (2730)	3500 (3230)	3900 (3630)	4200 (3930)	5000 (4730)
				HfC_{1-x}–C phases equilibrium					
lgP_{Hf}, Pa	−31.700	−16.913	−9.521	−5.075	−2.134	−0.054	1.212	2.004	3.407
lgP^{a}_{Σ}, Pa	−24.287	−11.719	−5.224	−1.156	1.595	3.564	4.771	5.455	6.394
lg(C/Hf)[b]	6.420	5.270	4.590	4.300	4.080	3.900	3.774	3.621	3.082
Contents, vol.%:									
C	99.89	89.76	50.56	23.59	12.76	8.00	5.96	5.48	9.62
C_2	0.01	0.82	4.04	6.68	8.16	8.94	9.19	9.70	15.52
C_3	0.10	9.42	45.38	69.47	77.42	77.15	72.87	68.54	59.27
C_4	–	–	–	0.04	0.26	0.88	1.77	2.53	3.62
C_5	–	–	0.01	0.20	1.38	4.99	10.15	13.67	11.72
Hf	–	–	0.005	0.012	0.019	0.024	0.028	0.036	0.103
HfC[c]	–	–	–	0.001	0.004	0.012	0.023	0.038	0.133

[a]Total gas pressure
[b]Logarithm of atomic ratio in gaseous phase
[c]Gaseous molecule

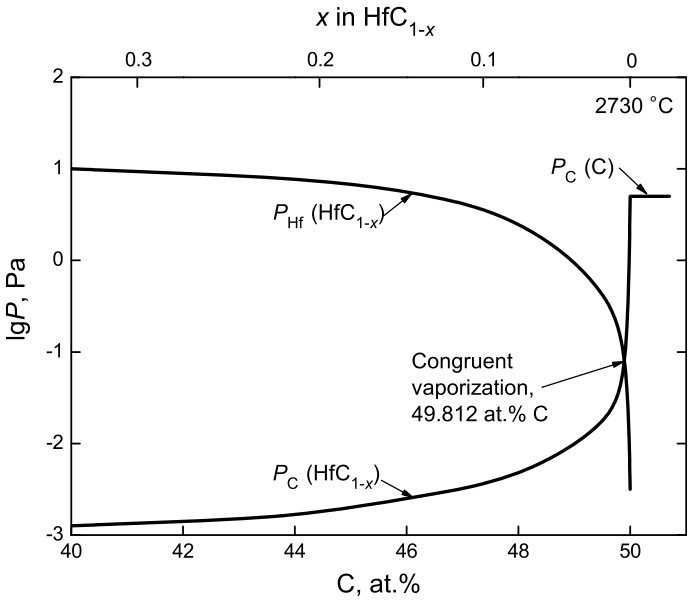

Fig. 3.4 Partial pressures of hafnium P_{Hf} (HfC_{1-x}) and carbon P_C (HfC_{1-x}) over hafnium carbide HfC_{1-x} phase, and vapour pressure of carbon over pure graphite P_C (C) as functions of carbide composition at 3000 K (2730 °C) calculated by Kaufman [9, 14, 15, 31, 296]; a congruently vaporizing composition is marked

$$\lg x = 3.149 - \left(1.372 \times 10^4\right)/T, \tag{3.23}$$

where x is the value of index in HfC_{1-x} formula and T is temperature, K, while for the rate of congruent vaporization of hafnium carbide materials from an open surface into a vacuum V, g cm^{-2} s^{-1}, for the range of 2580–3300 K (2310–3030 °C) the equation, which is similar to Eq. (3.22),

$$\lg V = 8.409 - \left(3.916 \times 10^4\right)/T, \tag{3.24}$$

was given, that is rather close to the results obtained by Fesenko and Bolgar [76] earlier.

The values of general thermodynamic properties, vapour pressures and vaporization rates for hafnium monocarbide are given in Addendum in comparison with other ultra-high temperature materials in the wide ranges of temperatures.

At room temperature thermal conductivity of near-stoichiometric hafnium monocarbide HfC_{1-x} is about 5–30 W m^{-1} K^{-1} (thermal diffusivity ~0.04 cm^2 s^{-1}) [1, 4, 32, 35, 36, 53, 61, 68, 77, 78, 82, 91, 154, 310, 378, 379, 525, 539]; it is affected by the porosity of carbide materials noticeably. Within the homogeneity range, as a consequence of conduction electrons scattering on the carbon sublattice vacancies and thermal lattice vibrations, the thermal conductivity of HfC_{1-x} declines

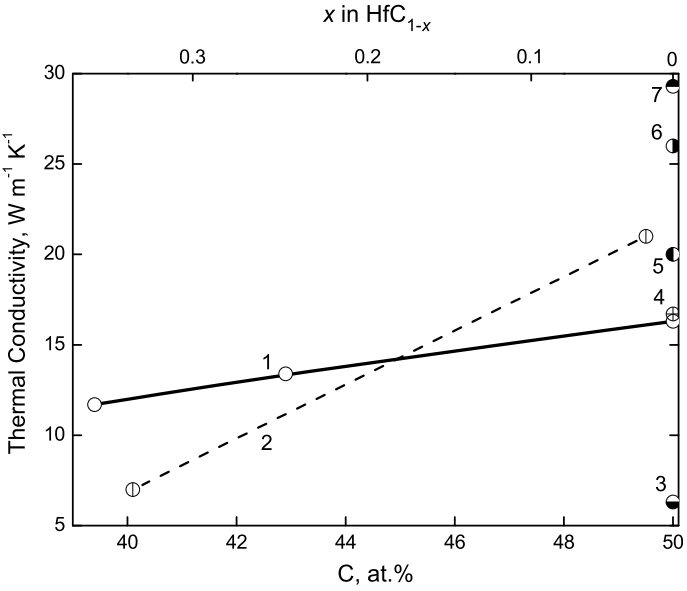

Fig. 3.5 Thermal conductivity of hafnium monocarbide HfC_{1-x} at room temperature (within the homogeneity range) as a function of carbide composition (1 – calculated on the basis of Goryachev's method [1]; 2 – hot-pressed, porosity 5–10% [91]; 3 – [4, 35, 36]; 4 – [53]; 5 – [61]; 6 – [68, 77]; 7 – [32])

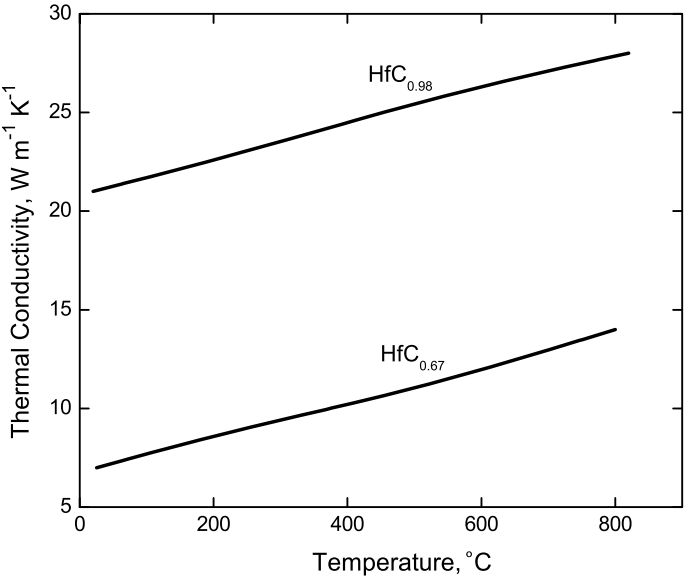

Fig. 3.6 Thermal conductivity of HfC_{1-x} as a function of temperature for the phases having different deviations from the stoichiometry [91]

with increasing carbon deficit (value of x) in the phase [14, 15] that is shown in Fig. 3.5; the difference in the temperature dependences of thermal conductivity for the HfC_{1-x} phases having different deviations from the stoichiometry is illustrated by Fig. 3.6. For the experimental data fit of thermal conductivity λ of $HfC_{0.96}N_{0.04}$ phase Lengauer et al. [29] obtained the following polynomial expression for the range of 300–1100 K (25–830 °C):

$$\lambda = 19.48 + (14.26 \times 10^{-3})T - (4.788 \times 10^{-6})T^2 - (4.614 \times 10^5)T^{-2}, \quad (3.25)$$

where T is temperature, K. The variation of thermal conductivity with temperature for near-stoichiometric HfC_{1-x} materials on the basis of several sources is shown in Fig. 3.7.

At room temperature the mean coefficient of linear thermal expansion of $HfC_{\sim 1.0}$ is (4.6–6.6) $\times 10^{-6}$ K^{-1} [34, 45, 61, 68, 81, 85, 139, 582]. The experimental data collected from the various measurements of thermal expansion of hafnium monocarbide are listed in Table 3.6. The approximation functions for the temperature

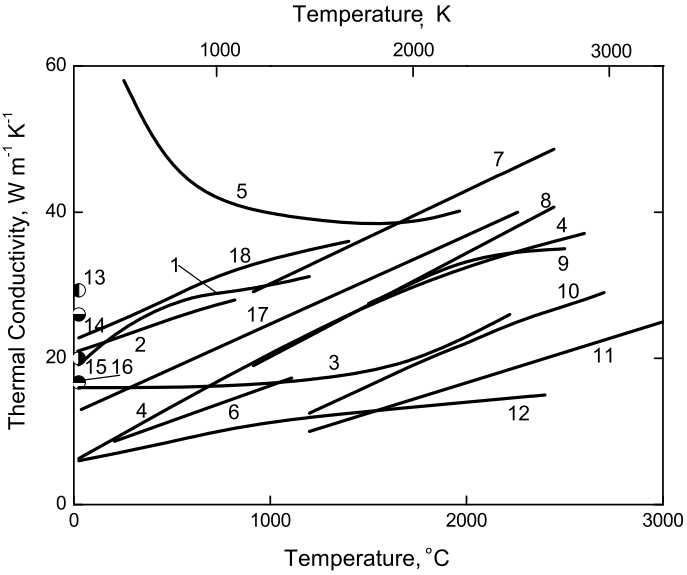

Fig. 3.7 Variation of thermal conductivity with temperature for near-stoichiometric hafnium monocarbide HfC_{1-x} on the basis of several sources: 1 – [378, 379]; 2 – hot-pressed $HfC_{0.98}$, porosity 5–10% [91]; 3 – [146]; 4 – chemically vapour deposited [4, 35, 36, 78]; 5 – hot-pressed and additionally heat-soaked at 2200 °C, porosity 6–12% [139]; 6 – hot-pressed, porosity 6–12% [139]; 7 – hot-pressed $HfC_{0.96}$, porosity 4–6%, contents: non-combined C – 0.03%, Zr – 3.45% [79, 307, 308]; 8 – hot-pressed $HfC_{0.96}$, porosity 4–6%, contents: non-combined C – 0.03%, Zr – 3.45% (repeated test) [79, 307, 308]; 9 – chemically vapour deposited [14, 15]; 10 – $HfC_{0.93}$, porosity 22–28% [81]; 11 – hot-pressed [14, 15]; 12 – hot-pressed, porosity 3% [310]; 13 – [32]; 14 – [68, 77]; 15 – [61]; 16 – [53]; 17 – [82]; 18 – spark-plasma sintered, porosity 15% [539]; when it is not indicated specially, data given are for quasi-stoichiometric composition

Table 3.6 Average coefficients of linear thermal expansion α_m of hafnium monocarbide HfC_{1-x} in various temperature ranges

Temperature range, °C	α_m, 10^{-6} K^{-1}	References
(−190)–(+25)	3.5 ± 0.4[a]	[20]
	4.1 ± 0.7[b]	[20]
	4.3 ± 0.4[c]	[20]
(–190)–(+20)	3.15[d]	[582]
20–200	6.0	[45, 53]
	6.22[d]	[582]
100–200	6.3[e]	[137]
20–230	6.39	[501, 502]
20–300	6.30[d]	[582]
100–300	6.4[e]	[137]
20–400	6.37[d]	[582]
	6.50	[45, 53]
20–500	6.32	[138]
	6.45[d]	[582]
100–500	6.5[e]	[137]
20–540	4.6	[139]
20–610	6.59 ± 0.04[f]	[85]
20–600	6.51[d]	[582]
	6.66	[45, 53]
20–650	6.30	[53]
20–700	6.58[d]	[582]
100–700	6.7[e]	[137]
20–730	6.64	[501, 502]
20–800	6.5	[45, 53]
	6.63[d]	[582]
20–815	6.3	[82]
20–900	5.1	[89]
	5.7–5.9	[77]
	6.68[d]	[582]
100–900	6.6 ± 0.2[e]	[137]
20–1000	6.19[g]	[87]
	6.26	[138]
	~6.35[h]	[91]
	6.6	[45, 53, 388]
	6.70 ± 0.08[i]	[84, 490]
	6.73[d]	[582]
	8.23	[4]
20–1020	5.0	[139]
20–1100	6.77[d]	[582]
20–1200	6.06	[88]
	6.10	[92, 378, 381]
	6.67	[45, 53]

(continued)

Table 3.6 (continued)

Temperature range, °C	α_m, 10^{-6} K^{-1}	References
	6.73	[385]
	6.78[j]	[389]
	6.82[d]	[582]
20–1300	6.85[d]	[582]
20–1330	6.89	[501, 502]
20–1400	6.80	[45, 53]
	6.89[d]	[582]
	7.00	[90]
20–1500	6.59[g]	[87]
	6.91[d]	[582]
20–1600	6.87	[45, 53]
	6.94[d]	[582]
20–1730	6.98	[501, 502]
	7.0	[31]
20–1800	6.66	[45, 53]
	6.97[d]	[582]
	7.47	[90]
20–2000	6.8	[45, 53]
	7.00[d]	[582]
	7.10[g]	[87]
	7.66 ± 0.11[k]	[539]
20–2025	7.14[l]	[34]
500–2050	7.13[m]	[582, 583]
20–2130	7.05	[501, 502]
20–2200	5.2	[67]
	6.9	[45, 53]
	7.06[d]	[582]
815–2200	9.0	[82]
20–2400	6.9	[45, 53]
	7.14[d]	[582]
	7.34[g]	[87]
20–2500	7.18[d]	[582]
20–2600	7.54[g]	[87]
	7.24[d]	[582]
20–2700	7.29[d]	[582]
	7.3 ± 0.2[n]	[86]
20–2800	7.34[d]	[582]
20–2900	7.40[d]	[582]
	8.01	[90]

(continued)

Table 3.6 (continued)

[a]Powdered $HfC_{0.90}$, measured by low-temperature X-ray diffraction method
[b]Powdered $HfC_{0.97}$, measured by low-temperature X-ray diffraction method
[c]Powdered $HfC_{0.95}$, measured by low-temperature X-ray diffraction method
[d]Calculated on the basis of approximation function
[e]Fused $HfC_{0.80}$ (contents: O – 0.05%, N – 0.003%, Zr – 2%), measured by dilatometric method
[f]Hot-pressed $HfC_{\sim 1.0}$ (1–4% porosity), measured by dilatometric method
[g]$HfC_{0.96}$ hot-pressed at 3100–3150 °C and measured by dilatometric method
[h]Hot-pressed $HfC_{0.67}$ and $HfC_{0.98}$ (5–10% porosity), measured by dilatometric method
[i]For $HfC_{\sim 1.0}$ (for $HfC_{0.67}$ $\alpha_m = 6.79 \pm 0.08\ 10^{-6}\ K^{-1}$)
[j]Measured by high-temperature X-ray diffraction
[k]Spark-plasma sintered $HfC_{\sim 1.0}$ (15% porosity, mean grain size – 5.8 ± 0.2 μm), measured by dilatometric method in Ar atmosphere
[l]Sintered $HfC_{0.95}$, measured by high-temperature X-ray diffraction
[m]Materials prepared by the diffusion saturation of metallic Hf by carbon, measured by telemicroscope method in He atmosphere
[n]$HfC_{\sim 1.0}$ grains in a matrix of the arc-cast carbide-carbon eutectic materials, measured by high-temperature X-ray diffraction with the usage of (111) reflections

dependence of relative thermal linear expansion $\Delta l/l_0 = f(T,\ \text{K})$, %, of near-stoichiometric HfC_{1-x} was recommended by Touloukian et al. [582] on the basis of experimental data for pure samples [20, 86, 139, 310, 583] (accuracy within ±5% at $T < 1800$ K and within ±10% at $T > 1800$ K):

for 293–2000 K (20–1730 °C):

$$\Delta l/l_0 = -0.170 + (5.521 \times 10^{-4})T + (1.041 \times 10^{-7})T^2 - (2.015 \times 10^{-11})T^3, \tag{3.26}$$

and for 2000–3200 K (1730–2930 °C):

$$\Delta l/l_0 = 7.367 \times 10^{-2} + (4.217 \times 10^{-4})T + (6.072 \times 10^{-8})T^2 + (3.374 \times 10^{-12})T^3, \tag{3.27}$$

where T is temperature, K. On the basis of high-temperature X-ray measurements Houska [34] proposed for the average coefficient of linear thermal expansion α_m, K^{-1}, of sintered (in vacuum) $HfC_{0.95}$ (contents: non-combined C – 2.04%, N – 0.08%, O – 0.06%) in the temperature range of 25–2025 °C the following equation:

$$\alpha_m = 6.38 \times 10^{-6} + (0.38 \times 10^{-9})(t - 25), \tag{3.28}$$

where t is temperature, °C. The extrapolation of the relationship, obtained by Aigner et al. [92] for the range of 25–1200 °C, to hypothetical $HfC_{1.0}$ phase gives the following expression for its coefficient of linear thermal expansion α, K^{-1}:

$$\alpha = \frac{2.2791 \times 10^{-6} + 0.494 \times 10^{-9}T}{0.46323 + 2.2791 \times 10^{-6}T + 0.247 \times 10^{-9}T^2}, \tag{3.29}$$

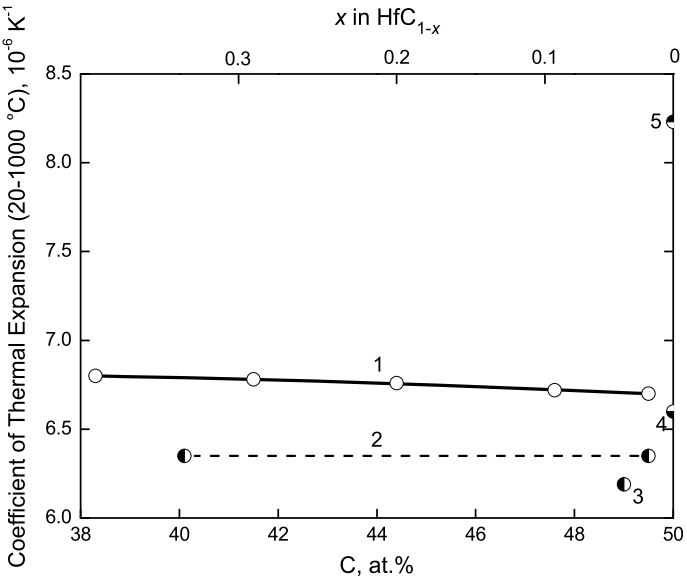

Fig. 3.8 Coefficient of linear thermal expansion in the temperature range of 20–1000 °C of hafnium monocarbide HfC_{1-x} (within the homogeneity range) materials as a function of carbide composition (1 – hot-pressed, 6–10% porosity, content non-combined C $\leq$ 0.16%, measured by dilatometric method [84, 480, 488]; 2 – hot-pressed, 5–10% porosity, measured by dilatometric method [91]; 3 – hot-pressed, measured by dilatometric method [87]; 4 – [45, 53]; 5 – [4])

where T is temperature, K. The influence of carbon content on the thermal expansion of HfC_{1-x} phase within the homogeneity range can be seen in Fig. 3.8; although according to Opeka et al. [91] there is no difference in thermal expansion behaviour measured by dilatometric method for hot-pressed $HfC_{0.67}$ and $HfC_{0.98}$ materials with 5–10% porosity. Some recents works [62, 318, 336, 391] are devoted to the calculations of coefficients of thermal expansion of hafnium monocarbide by means of theoretical modelling.

In comparison with other ultra-high temperature materials the values of thermal conductivity and thermal expansion of hafnium carbide in the wide range of temperatures are summarized in Addendum.

3.3 Electro-magnetic and Optical Properties

At room temperature the value of specific electrical resistance (resistivity) of near-stoichiometric hafnium monocarbide HfC_{1-x} lies within the area of 0.2–0.9 μΩ m [1, 14, 15, 31, 45, 53, 68, 84, 300, 390, 490, 499, 586]. Anomalously high resistivity with a negative thermal coefficient was reported by Lei et al. [494] for a $HfC_{0.88}$ thin film (thickness – 1.2 μm) prepared by evaporation technique.

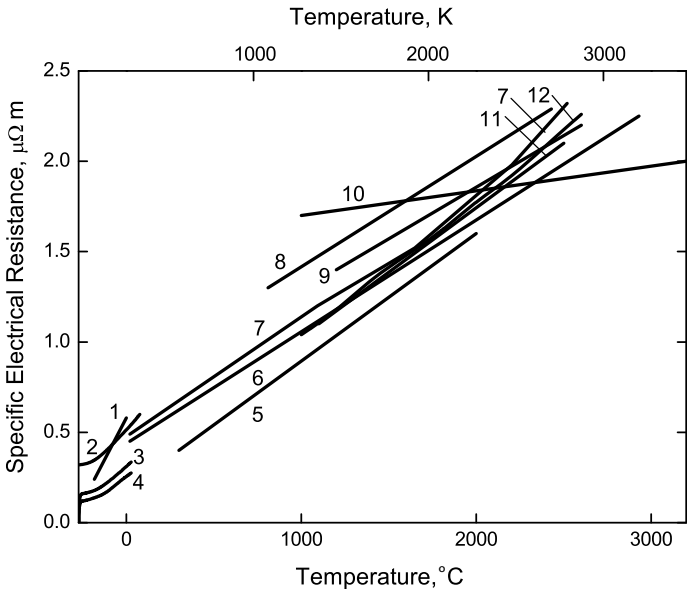

Fig. 3.9 Variation of specific electrical resistance (resistivity) with temperature for near-stoichiometric hafnium monocarbide HfC_{1-x} on the basis of several sources: 1 – sintered, 4% porosity [413]; 2 – single crystal $HfC_{0.99}$ [414]; 3 – hot isostatically pressed and subsequently sintered $HfC_{0.98}$, 26% porosity [93]; 4 – hot-pressed $HfC_{0.97}$, ~2.5% porosity [93]; 5 – [94]; 6 – counted for a pore-free state [31]; 7 – hot-pressed $HfC_{0.96}$, porosity 4%, contents: non-combined C – 0.03%, Zr – 3.45% and 8 – same (repeated test) [79, 307]; 9 – $HfC_{0.93}$, 22–28% porosity [81]; 10 – [95]; 11 – [45]; 12 – counted for a pore-free state [53]

The variation of specific electrical resistance with temperature for near-stoichiometric HfC_{1-x} is shown on the basis of several sources in Fig. 3.9. In the wide temperature range from 0–200 °C up to 3000–3300 °C the resistance of monocarbide phase enlarges with increasing temperature, practically in accordance with linear relationship; that is an evidence of mainly metallic type of conduction in HfC_{1-x} [14, 15]. The single crystal $HfC_{0.99}$ resistivity measured by Modine et al. [414] at ultra-low, low and ambient temperatures is perfectly described in the interval from $T = 0$ to $T = 350$ K by the Wilson expression (in μΩ m):

$$\rho(T) = 0.322 + (1.467 \times 10^{-8})T^3 \int_0^{352/T} x^3 / \sinh^2 x dx. \qquad (3.30)$$

The experimental data collected from the various measurements of thermal coefficient of resistivity for near-stoichiometric hafnium monocarbide are listed in Table 3.7. The data on the resistivity of HfC_{1-x} within the homogeneity range are contradictory; it was established that the resistivity enlarges with increasing carbon deficit in non-metal sublattice [84, 97], whereas in other work [96] the reverse

Table 3.7 Average values of thermal coefficients of resistivity α_R of near-stoichiometric hafnium monocarbide HfC_{1-x} in various temperature ranges

Temperature range, °C	α_R, 10^{-3} K^{-1}	References
(–195)–0	6.83[a]	[367]
(–183)–0	7.74[b]	[413]
(–123)–80	2.69[c]	[414]
20–1000	0.94[d]	[98]
	0.43[e]	[494]
20–1150	0.41	[53]
300–2000	1.42	[1, 31, 94, 478, 479]
20–2890	0.27	[53]

[a]Materials deposited from gas phase
[b]Sintered materials, 4% porosity
[c]Single crystal $HfC_{0.99}$, $d\rho/dT = 9.1 \times 10^{-4}$ μΩ m K^{-1}
[d]Hot-pressed, counted for a pore-free state, $d\rho/dT = 5.0 \times 10^{-4}$ μΩ m K^{-1}
[e]Thin films prepared by activated reactive evaporation, carbon content – 46.8 at.%, thickness – 1.2 μm

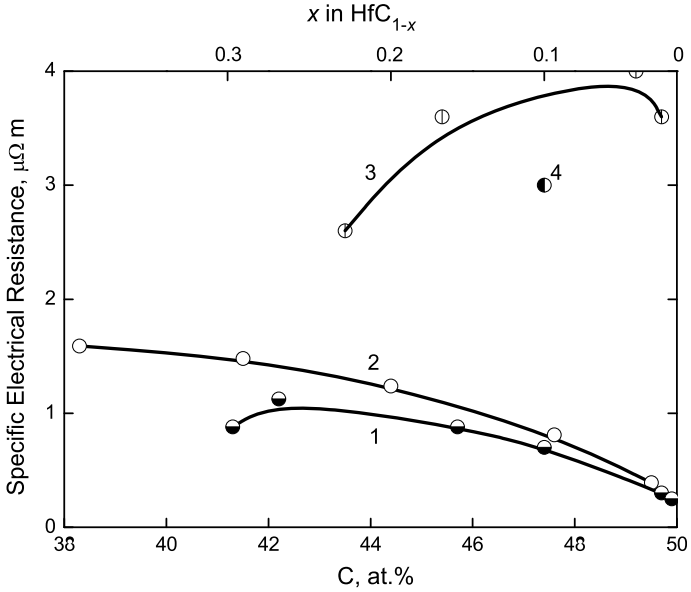

Fig. 3.10 Variation of specific electrical resistance (resistivity) at room temperature within the homogeneity range of hafnium monocarbide HfC_{1-x} on the basis of several sources: 1 – [97]; 2 – sintered, 15–25% porosity [84]; 3 – [96]; 4 – [415]

picture was observed that is demonstrated in Fig. 3.10. In contrast to some other refractory transition metal carbides hafnium monocarbide HfC_{1-x} is not a superconductor along its full homogeneity range (T_c < 1.23 K) [14, 61, 293, 294, 487, 511]. At room temperature the Hall and Seebeck coefficients of near-stoichiometric hafnium monocarbide HfC_{1-x} are $R = -(11–18) \times 10^{-10}$ m^3 A^{-1} s^{-1} and $S = -(3.5–12.5)$ μV K^{-1}, respectively [14, 15, 68, 96, 101, 132, 415–418]; experimental data obtained by Golikova et al. [96] and Borukhovich et al. [415–417] for the Seebeck coefficients within the homogeneity range of HfC_{1-x} are presented in Fig. 3.11.

Near-stoichiometric hafnium monocarbide HfC_{1-x} is a diamagnetic substance with molar magnetic susceptibility χ_m (SI) ≈ $-(200–500) \times 10^{-6}$ cm^3 mol^{-1} at room temperature slightly increasing with temperature in the range from 20 to 1000 °C [2, 5, 6, 9, 14–16, 100, 484, 509]. However, an increase in carbon deficit (value of x) in HfC_{1-x} results in changing in the magnetic properties of the phase from diamagnetic to paramagnetic, so the maximum of molar magnetic susceptibility χ_m (SI) ≈ $(275–530) \times 10^{-6}$ cm^3 mol^{-1} is corresponding to the lowest carbon content in the phase and χ_m (SI) ≈ 0 – to the composition of $HfC_{0.77–0.80}$ (Fig. 3.12). Gusev and Zyryanova [5, 6] proposed for HfC_{1-x} within its homogeneity range the equation

$$\chi_m = A + BT^2, \tag{3.31}$$

which describes the relationship between magnetic susceptibility and temperature T, K in the interval from room temperature to disorder-order or order-disorder transformation temperatures (~800 K) and corresponds to the temperature dependence

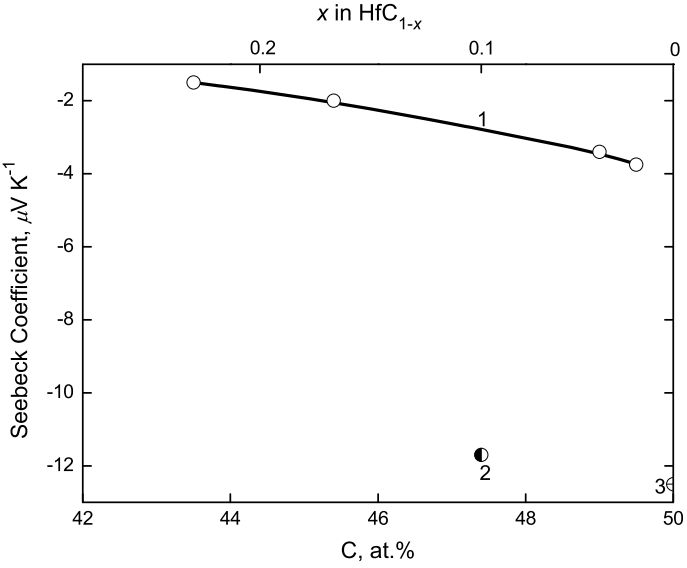

Fig. 3.11 Variation of Seebeck coefficient at room temperature within the homogeneity range of hafnium monocarbide HfC_{1-x} (1 – [96]; 2 – [415–417]; 3 – hot-pressed [101, 418])

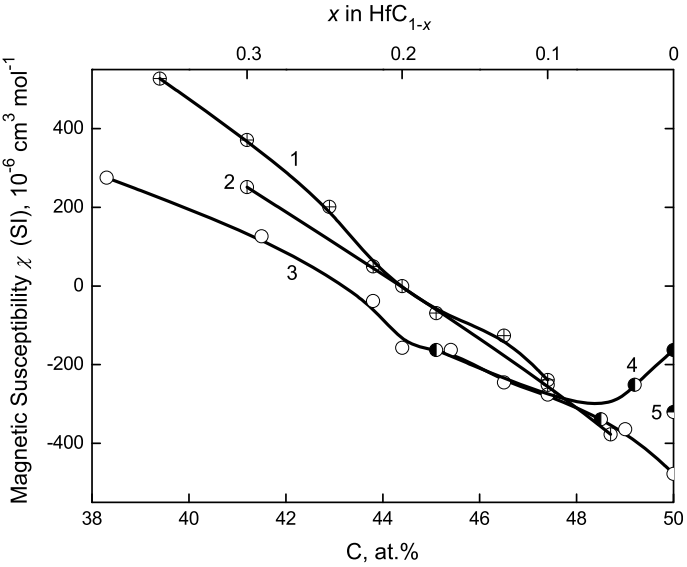

Fig. 3.12 Variation of molar magnetic susceptibility at room temperature within the homogeneity range of hafnium monocarbide HfC_{1-x} on the basis of several sources: 1 – [16], 2 – [1, 14, 15, 35], 3 – contents: non-combined C – 0.01–0.08%, O – 0.21–0.45%, N – 0.05–0.07% [5, 6, 484–486], 4 – [100], 5 – [77, 101]

of Pauli paramagnetism; where χ_m is the molar magnetic susceptibility (SI) of HfC_{1-x}, $cm^3\ mol^{-1}$, A is the empirical constant increasing from -353×10^{-6} $cm^3\ mol^{-1}$ (for $x = 0.04$) to 248×10^{-6} $cm^3\ mol^{-1}$ (for $x = 0.38$) and B is also the constant calculated from the experimental data (temperature coefficient) varying in the range of $(6620–10980) \times 10^{-14}$ $cm^3\ mol^{-1}\ K^{-2}$. The values of A and B in the function χ_m, which is sensitive to the ordering in the structure of HfC_{1-x} (*see* Sect. 3.1), change with disorder-order or order-disorder transformations [5, 6].

The room-temperature reflectance spectrum of pure spark plasma sintered HfC_{1-x} (porosity – 2%, grain size – ~7 μm) from the ultraviolet wavelength region to the mid-infrared band is shown in Fig. 3.13. The optical properties of quasi-stoichiometric HfC_{1-x} were reported by Delin et al. [588] on the basis of *ab initio* calculations, and IR absorption spectra – by Kammori et al. [589]. At common conditions the colour of hafnium monocarbide HfC_{1-x} in the dispersed state (powder) is grey or silver-gray [1, 35, 45, 61]. The normal monochromatic emittance (spectral emissivity) ε_λ of hafnium monocarbide slightly varies with temperature, the linear relationship for $\lambda = 0.65$ μm

$$\varepsilon_\lambda = 0.4322 + (1.065 \times 10^{-4})(t + 273), \tag{3.32}$$

where t is temperature, °C, was obtained by Mackie et al. [105, 378] for arc-melted (via floating zone) polycrystalline $HfC_{0.86\pm0.03}$ in the temperature range 1150–1950 °C, while Samsonov et al. [14] recommended linear equations, but with negative temperature coefficients:

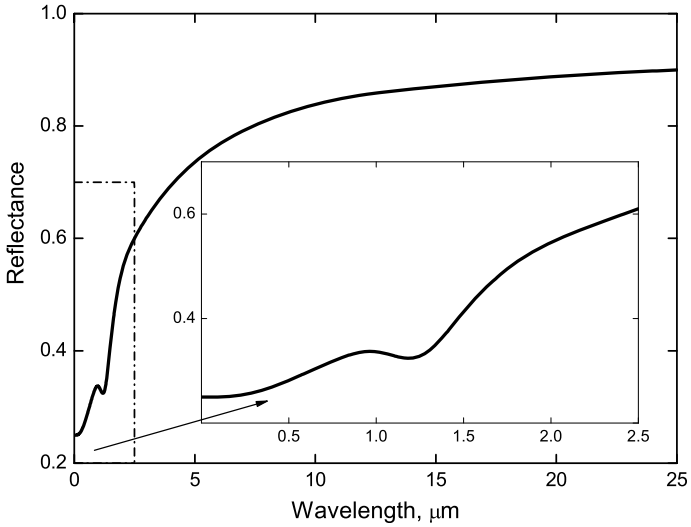

Fig. 3.13 Room-temperature reflectance spectrum of pure spark plasma sintered near-stoichiometric HfC_{1-x} (2% porosity, grain size ~7 μm) from the ultraviolet wavelength region to the mid-infrared band [102–104]

Table 3.8 Normal monochromatic emittance ε_λ (λ = 0.650–0.665 μm) of hafnium monocarbide HfC_{1-x} materials in various temperature ranges

Monochromatic emittance, ε_λ	Temperature range, °C	References
0.51–0.73[a]	1600–2700	[109]
0.55–0.70[b]	700–2200	[105, 378]
0.66–0.67[c]	2100–2200	[1, 106]
0.76[d]	2000–3000	[14]
0.77[e]	800–1600	[31, 32, 68, 107, 108, 421]
0.80[d]	1000–1700	[14]
0.90[f]	1400–1900	[419]

[a]Hot-pressed $HfC_{0.98}$ (5–15% porosity); ε_λ falls with the temperature growth linearly
[b]Arc-melted (via floating zone) polycrystalline $HfC_{0.86\pm0.03}$, measured in vacuum 10^{-5}–10^{-6} Pa; ε_λ increases with the temperature growth linearly (slope is ~1.1 × 10^{-4} K^{-1})
[c]Pyrolytic coating; ε_λ falls with the temperature growth
[d]Hot-pressed materials
[e]Powdered materials (coatings) on tantalum substrate
[f]For λ = 0.81 μm, the value is constant and rather independent on surface roughness and thermal treatment in vacuum and hot hydrogen

Table 3.9 Integral (total) emittance ε_T of near-stoichiometric hafnium monocarbide HfC_{1-x} in various temperature ranges

Integral emittance, ε_T	Temperature range, °C	References
0.34–0.60[a]	850–1150	[524]
0.415–0.48[b]	1000–2700	[1, 110]
0.42–0.485[c]	2100–2600	[106, 420]
0.60–0.61[d]	2000–3000	[14, 15]
0.60–0.85[e]	1000–2700	[111]
0.65[f]	3900	[420]

[a]Hemispherical emittance ε_T for hot-pressed $HfC_{\sim 1.0}$ (<4% porosity, mean grain size – ~1 μm, surface roughness R_a = 0.3 μm); ε_T increases with the temperature growth (addition of $MoSi_2$ increases ε_T – up to 0.53–0.72 in this temperature range)
[b]Normal emittance ε_T for sintered $HfC_{\sim 1.0}$ (9% porosity, contents: non-combined C – 1.16%, O – 0.1%); ε_T increases with the temperature growth linearly
[c]Normal emittance ε_T for pyrolytic coating; ε_T increases with the temperature growth
[d]Normal emittance ε_T for sintered materials
[e]Normal emittance ε_T; curve $\varepsilon_T - T$ with the maximum at ~1300 °C and minimum at ~2400 °C
[f]Theoretically calculated normal emittance ε_T for the stoichiometric composition

$$\varepsilon_\lambda = 0.73 - (3.75 \times 10^{-6})(t - 727), \tag{3.33}$$

$$\varepsilon_\lambda = 0.89 - (1.65 \times 10^{-4})(t - 727), \tag{3.34}$$

where t is temperature, °C and which are aimed for 730–1600 and 1300–2500 °C intervals, respectively. The data on monochromatic emittance ε_λ (λ = 0.650–0.665 μm) and integral (total) emittance ε_T for HfC_{1-x} materials produced by different manufacturing methods and measured at various temperatures are listed in

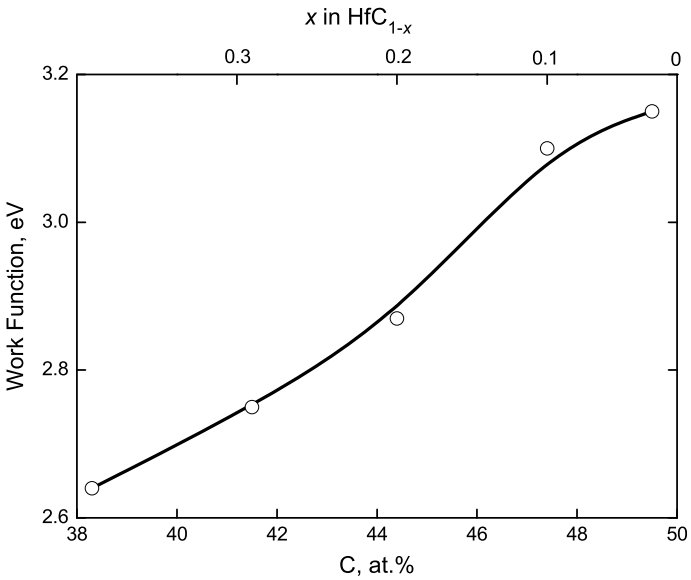

Fig. 3.14 Variation of effective work function of hot-pressed hafnium monocarbide HfC_{1-x} materials at 1730 °C with carbon content [112, 113, 488]

Table 3.10 Thermionic emission characteristics (electron work function and Richardson constant) of hafnium monocarbide HfC_{1-x} phases

Composition	Work function[a], $\varphi = \varphi_0 + (d\varphi/dT)_{av}T$, eV	Richardson constant, A, 10^4 A m^{-2} K^{-2}	Temperature range, °C	Remarks[b]	References
$HfC_{0.60}$	3.35	–	–	Theoretically evaluated	[498, 550, 551]
	3.87[c]	–	–	PhES, single crystal (100)	[497, 498, 550, 551]
$HfC_{0.62}$	$2.52 + (6.0 \times 10^{-4})T$	–	1200–1700	TiM, hot-pressed	[1, 113, 590–592]
	3.60[d]	–	1500	TiM, hot-pressed	[14, 15, 590–592]
$HfC_{0.67}$	3.66 ± 0.05	–	–	TiM, hot-pressed	[490]
$HfC_{0.71}$	$2.63 + (6.0 \times 10^{-4})T$	–	1200–1700	TiM, hot-pressed	[1, 113]
$HfC_{0.80}$	$2.75 + (6.0 \times 10^{-4})T$	–	1200–1750	TiM, hot-pressed	[1, 113]
$HfC_{0.88}$	3.32[d]	–	1500	FEM, single crystal (100)	[297]
$HfC_{0.90}$	$3.00 + (5.0 \times 10^{-4})T$	–	1200–1750	TiM, hot-pressed	[1, 113]
$HfC_{0.95}$	4.15[d]	–	1500	TiM, hot-pressed	[1, 114]
$HfC_{0.98}$	$3.05 + (5.0 \times 10^{-4})T$	–	1200–1700	TiM, hot-pressed	[1, 113, 590–592]
	3.95[d]	–	1500	TiM, hot-pressed	[14, 15, 590–592]
$HfC_{\sim 1.0}$	~2.04	10^{-5}	–	TiM, powder	[68, 122]
	2.94	–	–	FEM, thin films on W substrate	[595]
	$2.98 + (5.0 \times 10^{-4})T$	–	–	TiM, powder	[1, 115]
	2.99	–	–	FEM, thin films on Mo substrate	[595]
	3.00	0.3	–	TiM, powder on Ta foil	[1, 116]
	3.00	1.8	–	TiM, powder on Ta foil with $MoSi_2$ underlayer	[1, 116]
	3.25[d]	–	1300	TiM, powder	[1, 117]
	3.40	–	–	TiM, pyrolytic coating	[1, 118]
	$3.42 + (1.75 \times 10^{-4})T$	–	1000–1900	TiM, powder	[1, 14, 15, 113]
	3.46[d]	228	2300	TiM, powder on Pt substrate	[123, 594]
	3.57[d]	107.9	2300	TiM, metal-carburized wire	[123, 594]
	3.6 ± 0.1	~15–30	–	TiM, sintered	[1, 113]
	3.63 ± 0.05	20	–	TiM, pressed	[1, 113]
	3.66[d]	–	2100	TiM, hot-pressed	[1, 113]
	3.75[e]	–	–	Single crystal (001), theoretically evaluated	[327]
	3.80 ± 0.04[d]	–	1300	TiM, hot-pressed	[1, 119]
	3.85[d]	–	1900	TiM, powder on W substrate	[1, 120]
	3.86	–	–	Calculated value (unrelaxed surface)	[550]

(continued)

Table 3.10 (continued)

Composition	Work function[a], $\varphi = \varphi_0 + (d\varphi/dT)_{av}T$, eV	Richardson constant, A, 10^4 A m^{-2} K^{-2}	Temperature range, °C	Remarks[b]	References
	3.93[d]	–	1650	TiM, sintered	[1, 113]
	3.94[d]	39.8	2300	TiM, tip formed by electrolytic etching	[123, 594]
	4.0	40	1000–1800	TiM, massive disk cathode	[1, 121]
	4.00 ± 0.05	–	–	TiM, hot-pressed	[490]
	4.02[f]	–	–	Single crystal (001), theoretically evaluated	[327]
	4.10	–	2100	–	[50]
	4.15[d]	–	1500	TiM, sintered	[1, 14, 15, 113, 114]
	4.28	–	–	Calculated value (relaxed surface)	[550]
	4.31[g]	–	–	Single crystal (001), theoretically evaluated	[498, 550, 551]
	4.45[h]	–	–	Single crystal (100), theoretically evaluated	[505, 550, 551]
	4.50[i]	–	25	FEM, single crystal (001), clean surface	[380, 550, 551]
	4.5[j]	–	–	PhES, nanocrystalline thin films	[593]
	4.52	–	–	FEM, thin films on W substrate, clean surface	[595]
	4.60	–	–	FEM, thin films on Mo substrate, clean surface	[595]
	4.63[c]	–	–	PhES, single crystal (100)	[497, 498, 550, 551]
	4.90[k]	–	25	FEM, single crystal (111), clean surface	[380]

[a] T is temperature, K
[b] Methods applied for the experimental determination of the work function (TiM – thermionic method, PhES – photoemission spectroscopy (with synchrotron radiation), FEM – field emission microscopy) and manufacturing methods for the fabrication of a particular material (or its constitution) are marked
[c] Absolute work functions
[d] Values of effective electron work function
[e] Calculated using the Perdew's generalized gradient approximation (GGA) for the fixed surface model
[f] Calculated using the Perdew's GGA for the relaxed surface model
[g] Full-potential linear-muffin-tin-orbital (FP-LMTO) calculation technique (40-atom cell, no vacancies)
[h] Linear-muffin-tin-orbital with atomic sphere approximation (LMTO-ASA) calculation technique
[i] For graphene-covered surface φ = 4.3 eV
[j] In the presence of considerable amonts of amorphous C (ratio C/Hf ≈ 1.7)
[k] For graphene-covered surface φ = 3.7 eV

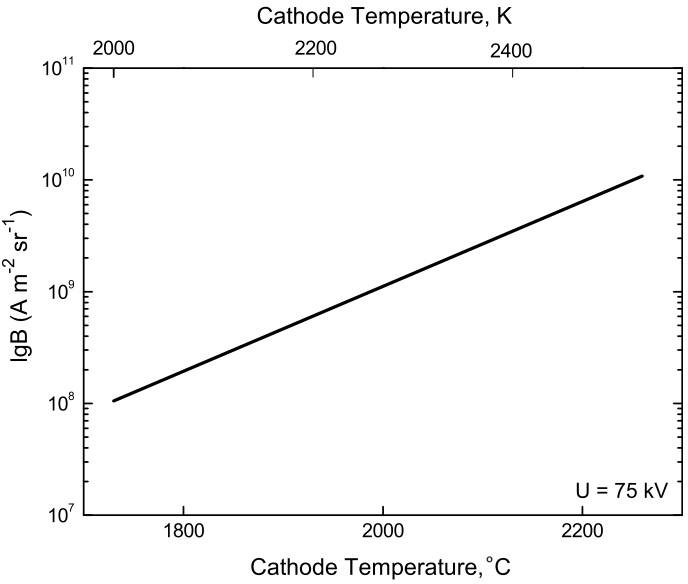

Fig. 3.15 Typical brightness (B) characteristic (U = 75 kV) of hafnium monocarbide cathode (thermionic emitter with electron work functions φ = 3.69 eV, φ_{eff} = 3.94 eV, Richardson constant A = 39.8 × 10^4 A m^{-2} K^{-2}, T = 2575 K) produced by carburization reaction of metal wire/foil with benzene vapour [123]

Tables 3.8 and 3.9. Thermodynamics of thermal radiation for stoichiometric hafnium monocarbide was considered by Fisenko and Lemberg [420].

The thermionic emission characteristics (electron work function and Richardson constants) of various hafnium monocarbide HfC_{1-x} phases are given in Fig. 3.14 and Table 3.10. The calculated values of emission current density for near-stoichiometric monocarbide HfC_{1-x} are about 3 × 10^{-4}, 0.3 and 50 A m^{-2} at 730 °C, 1230 °C and 1730 °C, respectively [31]. Similar to some other transition metal carbides, single crystal HfC_{1-x} materials show following relationships between the effective thermionic work functions of different crystallographic planes: $\varphi_{(210,310)} < \varphi_{(100)} \approx \varphi_{(110)} \gg \varphi_{(111)}$ [297]. A typical brightness characteristic (U = 75 kV) of hafnium monocarbide cathode (thermionic emitter for electron microscopy) is shown in Fig. 3.15. Field emission properties of HfC_{1-x} single crystal tips and thin films (coatings) were reported in several works [297, 593, 595–616].

The recommended values of electrical resistivity, magnetic susceptibility, integral and spectral emittances and thermionic emission characteristics (electron work function and Richardson constant) for hafnium monocarbide materials are given in the wide range of temperatures in comparison with other ultra-high temperature materials in Addendum.

3.4 Physico-mechanical Properties

The physico-mechanical properties of hafnium monocarbide HfC_{1-x} are sensitive to the deviations from stoichiometry, but also to the crystallographic directions in the materials. At room temperature the hardness *HV*, GPa of near-stoichiometric hafnium monocarbide HfC_{1-x} is evaluated as 10.2 ± 0.7 (porosity – 15%, mean grain size – 1.8 ± 0.2 μm, 9.8 N load) [539], 16.4 (for $HfC_{0.95}$) [425], 21.0 (contents: non-combined C – 0.09%, O – 0.09%, N – 0.05%, porosity – 5%, mean grain size – 3 μm, 1 N load) [427, 428], 22.6 [380], 24.8–31.4 [50, 154], 25.4 (0.5 N load) [77], 25.9 ± 1.6 [14, 15, 133], 26.0 ± 1.0 (contents: O + N – 0.1%, 0.5 N load) [131, 428], 26.1 [61], 27.0 ± 1.0 (for $HfC_{0.93}$, 1 N load) [4, 131], 27.8 ± 1.4 (for $HfC_{0.98}$, 0.5 N load) [35, 423], 28.1 ± 3.3 [129], 28.5 ± 3.0 (for monocarbide phase in contact with carbon, 0.5 N load) [31, 32, 424] and 29.0 (contents: O + N – 0.35%) [130]; hardness *HK*, GPa is 19.0–22.6 (0.5 N load) [422], 22.3 (extrapolated to $HfC_{1.0}$, 1 N load) [135] and 25.5 (for hafnium monocarbide HfC_{1-x} phase in contact with graphite, 1 N load) [17]. The hardness of near-stoichiometric hafnium monocarbide in Rockwell scale *HRA* is 84 kgf mm^{-2} (0.82 GPa) [4, 35, 129]. According to Avarbe et al. [426], the hardness *HV*, GPa of HfC_{1-x} (probably with noticeable oxygen contamination) is related with composition by the following equation (slightly modified):

$$HV = 19.6 + 0.51 \times (1 - x)/(2 - x), \tag{3.35}$$

where x is the value of index in HfC_{1-x} formula. According to Adams and Beall [135] the hardness *HK* of fused HfC_{1-x} materials in the interval between 50 and 30 at.% C changes with composition significantly and varies from 22.3 (value extrapolated to defectless theoretical stoichiometry) to 17.8 GPa (1 N load). The variations of hardnesses *HV* and *HK* with carbon content within the homogeneity range of HfC_{1-x} and indenter orientation for the (001) surface of $HfC_{0.98}$ single crystal are demonstrated in Fig. 3.16. The variation of the hardness *HV* of near-stoichiometric hafnium monocarbide HfC_{1-x} with temperature is presented in Fig. 3.17. In the range from 0 to 900 °C the hardness-temperature relationship for HfC_{1-x} materials can be described by the exponential equation [31, 432–434]:

$$HV = 29.4 \exp[-(14.7 \times 10^{-4})t], \tag{3.36}$$

or such approximate linear equation (for temperatures <500 °C) [560] as

$$HV = 34.4 - (2.03 \times 10^{-3})t, \tag{3.37}$$

where *HV* is Vickers hardness, GPa and t is temperature, °C. The data on hardness *HK* of hafnium monocarbide single crystals and microhardness $H\mu$ of various HfC_{1-x} compositions are listed in Tables 3.11 and 3.12.

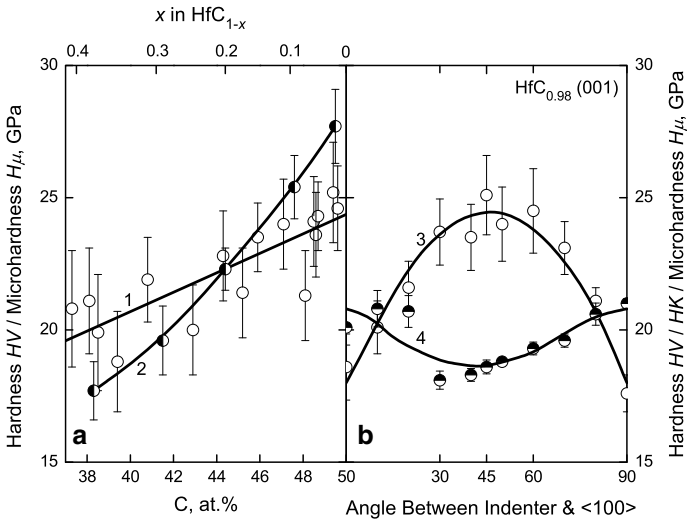

Fig. 3.16 Variations of the hardnesses *HV* (1, 4), *HK* (3) and microhardness $H\mu$ (2) at room temperature for hafnium monocarbide HfC_{1-x} materials: **(a)** with deviation from the stoichiometry within the homogeneity range (1 – prepared by various methods, 200 N load [21, 124, 125]; 2 – hot-pressed, 0.5 N load [4, 35, 84, 488]) and **(b)** with indenter orientation (angle between indenter axis and <100> direction) on the (001) surface of single crystal materials grown by floating-zone processes (3 – $HfC_{0.98}$, 1 N load and 4 – $HfC_{0.98}$, 2 N load [41, 42, 493])

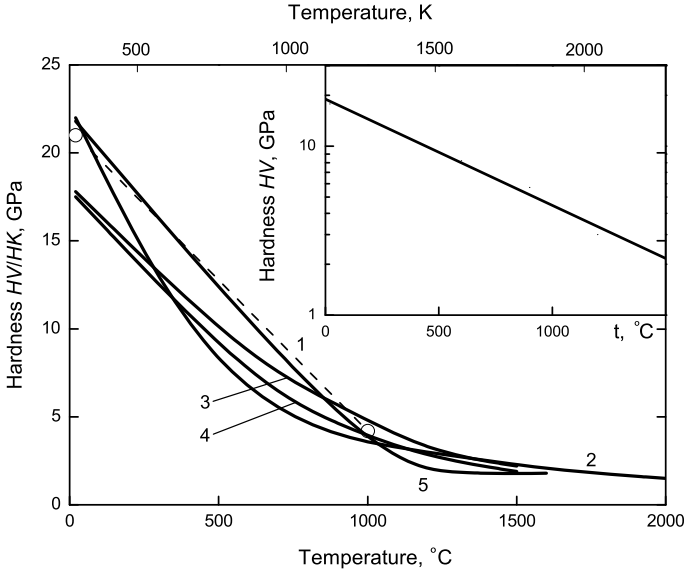

Fig. 3.17 Variations of the hardness *HV*/*HK* of hafnium monocarbide HfC_{1-x} with temperature: 1 – $HfC_{\sim 1.0}$, sintered in vacuum and annealed (mean grain size – 3.0 μm, porosity – 5%, contents: non-combined C – 0.09%, O – 0.09%, N – 0.05%), 1 N (microhardness) load [1]; 2 – $HfC_{\sim 1.0}$, hot-pressed and annealed in Ar atmosphere (porosity – 1%), 10 N load [126, 127]; 3 – $HfC_{0.87}$, fused (contents: O – 0.05%, N – 0.003%, Zr – 2%), 19.6 N load [137]; 4 – $HfC_{0.83}$, fused (contents: O – 0.05%, N – 0.003%, Zr – 2%), 19.6 N load [137]; 5 – $HfC_{\sim 1.0}$, zone-melted with excess C (~5%), 11 N load [364] (*Inset* – fused $HfC_{0.87}$, in lg(*HV*) – t, °C scale, 19.6 N load [137])

Table 3.11 Microhardness $H\mu$ (*HK* scale) of hafnium monocarbide HfC_{1-x} ($0 < x \leq 0.02$) single crystals at room temperature (1 N load)

Surface	Indentor diagonal direction	Microhardness $H\mu$, GPa	References
		$x \approx 0$[a]	
(001)	<100>	18.1	[617, 618]
	<110>	24.5	[617, 618]
	–	26.0 ± 0.5	[128]
	–	25.1 ± 0.5[b]	[128, 335]
(110)	–	22.9 ± 0.5	[128]
	–	22.7 ± 0.5[b]	[128, 335]
(111)	–	24.1 ± 0.5	[128]
	–	23.7 ± 0.5[b]	[128, 335]
		$x = 0.02$	
(001)	<100>	19.0 ± 1.0	[41]

[a]Crystals grown by controlled solidification from the melt
[b]Annealed at 2100 °C

Table 3.12 Microhardness $H\mu$ of the various compositions of hafnium monocarbide HfC_{1-x} phase at room temperature

Composition	Microhardness $H\mu$, GPa	Remarks	References
$HfC_{0.62}$	17.7 ± 1.1	0.5 N load, sintered in vacuum, contents: O – 0.09%, N – 0.04%	[35, 84]
$HfC_{0.63}$	19.5[a]	0.98 N load, fused, content non-combined C – ~0.05%	[135]
	15.8[a]	19.6 N load, fused, non-combined C – ~0.05%	[135]
$HfC_{0.67}$	18.8 ± 1.2	0.5 N load, sintered in vacuum	[84, 490]
$HfC_{0.71}$	19.6 ± 1.3	0.5 N load, sintered in vacuum, contents: O – 0.09%, N – 0.03%	[35, 84]
$HfC_{0.80}$	22.3 ± 0.9	0.5 N load, sintered in vacuum, contents: O – 0.10%, N – 0.06%	[35, 84]
$HfC_{0.82}$	21.0[a]	0.98 N load, fused, non-combined C – 0.04%	[135]
	18.2[a]	19.6 N load, fused, non-combined C – 0.04%	[135]
$HfC_{0.83}$	17.5 ± 0.9[b]	19.6 N load, fused, contents: O – 0.05%, N – 0.003%, Zr – 2%	[137]
$HfC_{0.87}$	17.8 ± 0.6[b]	19.6 N load, fused, contents: O – 0.05%, N – 0.003%, Zr – 2%	[137]
$HfC_{0.91}$	25.4 ± 1.2	0.5 N load, sintered in vacuum, contents: O – 0.09%, N – 0.04%	[35, 84]
	19 (22[a])	Electron beam deposited thin films on Ti substrate, thickness – 0.50 ± 0.03 μm, mean grain size <0.3 μm	[148]

(continued)

Table 3.12 (continued)

Composition	Microhardness $H\mu$, GPa	Remarks	References
$HfC_{0.95}$	25.9 ± 1.6	0.2–1.5 N load, hot-pressed and annealed at 1800 °C for 3 h, porosity – 7%, non-combined C – 0.1%	[129, 298]
	16.4[c]	–	[425]
$HfC_{0.96}$	28.8[d]	Hot-pressed, mean grain size – 5–20 μm	[441]
	25.6	Hot-pressed, mean grain size – 5–20 μm	[441]
	24.5[a]	0.5 N load, hot-pressed, porosity – 3–5%, mean grain size – 30 μm, contents: non-combined C – 0.03%, Zr – 3.45	[79, 80]
$HfC_{0.97}$	20.2[a]	0.5 N load, hot-pressed, porosity – 1–2%, mean grain size – 13 μm, contents: non-combined C – 0.30%, O – 0.45%, N – 0.10%, Zr – 1.40	[79, 80]
	20.0[a]	0.5 N load, sintered in vacuum, porosity – 4–7%, mean grain size – 42 μm, contents: non-combined C – 0.30%, O – 0.45%, N – 0.10%, Zr – 1.40	[79, 80]
$HfC_{0.98}$	27.7 ± 1.4	0.5 N load, sintered in vacuum, contents: non-combined C – 0.16%, O – 0.11%, N – 0.04%	[35, 84]
	27.4 ± 0.4	Pyrolytic materials	[430]
	25–28[a]	1 N load, sintered, porosity – 1–7%	[17]
$HfC_{\sim 1.0}$	29.0	Sintered in Ar, contents: O + N – 0.35%	[130, 299]
	28.5	–	[32, 108]
	28.4 ± 1.4	0.5 N load, sintered in vacuum	[84, 490]
	28.0	–	[14, 15]
	27.9	C_2H_2-reactive magnetron sputtered coatings	[321]
	27.7	1 N load, sintered	[64, 283]
	27.0	–	[9, 387]
	26.8	Calculated on the basis of Tian's model	[302]
	26.4 ± 1.6	–	[133]
	26.1	–	[61, 301]
	26.05[a]	0.98 N load, hot-pressed, porosity – 2%	[136]
	26.0 ± 1.0	Sintered, contents: O + N – 0.1%	[131, 363]
	25.7	Calculated value	[429]
	25.0	–	[129]
	24.0	–	[134]
	23.0[d]	0.5 N peak load, spark plasma sintered materials, porosity – 2%, mean grain size – 19 ± 1 μm	[103, 104, 435]
	22.8[a]	0.98 N load, fused, non-combined C – 0.09%	[135]
	21.8[a]	0.98 N load, zone-melted with excess C (~5%)	[364]
	21.0	Sintered in vacuum and annealed, porosity – 5%, mean grain size – 3.0 μm, contents: non-combined C – 0.09%, O – 0.09%, N – 0.05%	[1]
	20.1[e]	0.2–10 N load, hot-pressed, porosity <5%	[371]
	19.4–23.0[a]	0.5 N load	[67]

(continued)

Table 3.12 (continued)

Composition	Microhardness $H\mu$, GPa	Remarks	References
	19.4[a]	19.6 N load, fused, content non-combined C – 0.09%	[135]
	18.8 ± 0.7	9.8 N load, spark plasma sintered materials, porosity – 2%, mean grain size – 19 ± 1 μm	[103, 104, 435]
	18 ± 2	Pulsed laser ablation deposited thin films on Ti substrate, thickness – 0.22 to 0.50 μm, approximate grain size – 10 to 100 nm	[431]
	17.3	Theoretical evaluation	[305]
	17.9	–	[304]
	15.9 ± 0.5	0.1 kN load, reactive spark plasma sintered materials, porosity – 8%, mean grain size <5 μm	[436]
	15.6	Calculated on the basis of Xue's model	[302, 303]

[a]Knoop measurement (*HK*)
[b]Marshall micro-indentation diamond pyramid hardness testing in vacuum
[c]Diamond pyramid hardness measurement (*DPH*)
[d]Evaluated by a MTS XP diamond Berkovitch nanoindenter
[e]Determined in accordance with the formula: $H\mu = 18.18\ P/d^2$, where P is the load on the indentor, N and d is the diagonal of the indentation, μm; the values of other micromechanical characteristics: microbrittleness $\gamma_\mu = (D^2 - d^2)/d^2 = 3.0$, where D is the average size of the damageability zone, μm and brittle microstrength $\sigma_\mu = P/D^2 = 2.7$ GPa [371]

The variations of ultimate tensile and flexural (bending) strengths of near-stoichiometric hafnium monocarbide HfC_{1-x} with temperature determined in various experimental works are shown in Fig. 3.18 in comparison with the values of *ab initio* calculated strength characteristics. Spark-plasma sintered near-stoichiometric HfC_{1-x} (porosity 15%, mean grain size – 1.8 ± 0.2 μm) exhibits average fracture toughness K_{IC} of 2.9 ± 0.5 MPa $m^{1/2}$ [539], and for cold-pressed and sintered near-stoichiometric HfC_{1-x} (porosity 3–4%, mean grain size – 2.8 ± 1.0 μm, content O – 0.02%) $K_{IC} = 1.73 \pm 0.09$ MPa $m^{1/2}$ [149]. Brizes (cited by Hollox [438]) has reported that pure near-stoichiometric HfC_{1-x} materials are ductile at temperatures around 1600 °C. The values of prolonged strength (creep resistance) are of great importance for the evaluation of operational time of components in the engineering practice. The examples of ultra-high temperature creep characteristics of sintered and annealed in vacuum $HfC_{0.91}$ (porosity ≤ 10%, mean grain size – 22 μm, contents O + N – 0.24%) are demonstrated in Fig. 3.19; creep activation energies Q and exponent constants n of HfC_{1-x} phases are listed in Table 3.13. At room temperature the values of Young's modulus E of near-stoichiometric hafnium monocarbide HfC_{1-x} materials lies within the area of 320–480 GPa [1, 3, 4, 9, 79, 82, 91, 150, 311, 314]; the reported values for

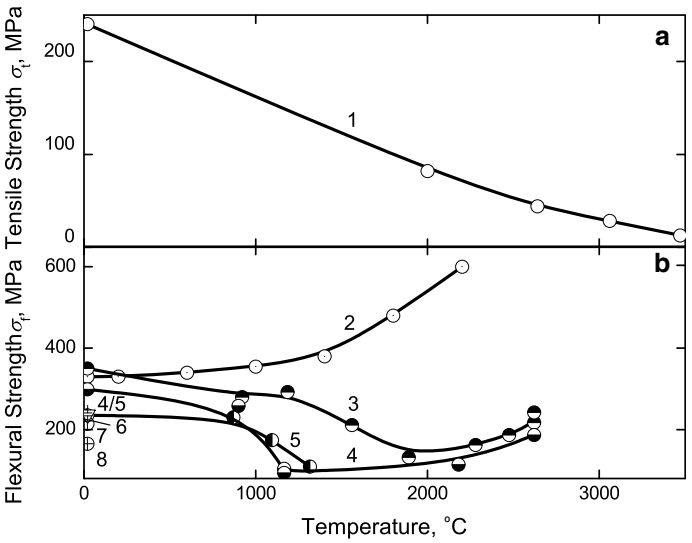

Fig. 3.18 Variations of (**a**) tensile σ_t (1 – sintered $HfC_{\sim 1.0}$, data corrected to the poreless state [14, 53, 88, 140]) and (**b**) flexural (bending) σ_f (2 – sintered $HfC_{\sim 1.0}$, data corrected to the poreless state [1, 141, 149]; 3 – hot-pressed $HfC_{\sim 1.0}$, porosity – 5%, mean grain size – 8 μm, contents: non-combined C – 0.04%, O – 0.16%, N – 0.03%, Zr – 1.07, 3-point bending scheme [80, 306]; 4 – hot-pressed $HfC_{\sim 1.0}$, porosity – 4%, mean grain size – 24 μm, contents: non-combined C – 0.04%, O – 0.16%, N – 0.03%, Zr – 1.07, 3-point bending scheme [80, 306]; 5 – [82, 312, 313]; 6 – hot-pressed $HfC_{0.96}$, porosity – 3–5%, mean grain size – 30 μm, contents: non-combined C – 0.03%, Zr – 3.45, 3-point bending scheme [79]; 7 – hot-pressed $HfC_{0.97}$, porosity – 1–2%, mean grain size – 13 μm, contents: non-combined C – 0.30%, O – 0.45%, N – 0.10%, Zr – 1.40, 4-point bending scheme [79]; 8 – sintered in vacuum $HfC_{0.97}$, porosity – 4 to 7%, mean grain size – 42 μm, contents: non-combined C – 0.30%, O – 0.45%, N – 0.10%, Zr – 1.40, 4-point bending scheme [79]) strengths with temperature for near-stoichiometric hafnium monocarbide HfC_{1-x} materials (the values of *ab initio* calculated ideal shear strength and ratio of the ideal strength to the elastic shear modulus for quasi-stoichiometric HfC_{1-x} are 30 GPa and ~0.13, respectively [362])

Coulomb's (shear) modulus G of HfC_{1-x} are 193–195 GPa [9, 150], 193–199 GPa [3, 4] (determined experimentally) and 166 GPa [62], 175–189 GPa [163], 230 GPa [164], 236 GPa [362] (calculated theoretically), for bulk (compression) modulus K – 245–290 GPa [3, 4, 9, 150, 152], for volume compressibility κ – 4.06–4.78 TPa^{-1} [1, 3, 4, 150, 151] and Poisson's ratio ν – 0.17 to 0.22 [3, 4, 150, 154, 163]; the *ab initio* calculated sound velocities (at zero pressure) for near-stoichiometric hafnium monocarbide $HfC_{\sim 1.0}$ are:

longitudinal velocity V_S, m s^{-1} 6964 (0 K) [318], 5474 (~300 K) [318], 6253 [315], 6530 [163], 6280–6450 [336], 6351 [444];

transversal velocity V_T, m s^{-1} 4273 (0 K) [318], 3339 (~300 K) [318], 3893 [315], 4090 [163], 3850–3950 [336], 3816 [444];

average velocity V_m, m s^{-1} 4250–4360 [336], 4880 [444].

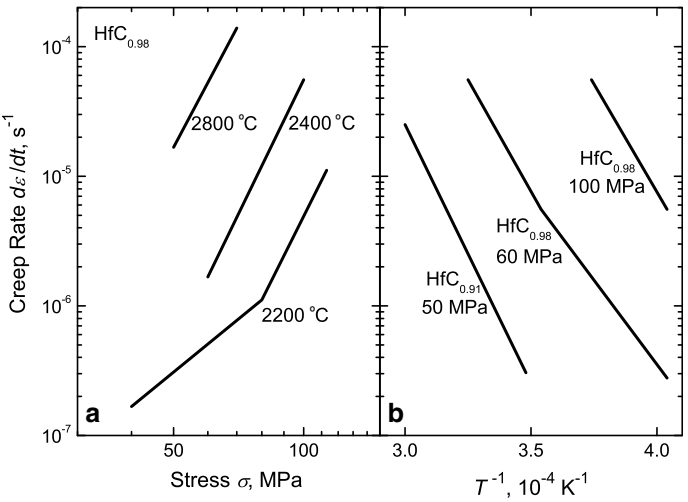

Fig. 3.19 High- and ultra-high temperature creep rate variations of near-stoichiometric hafnium monocarbide HfC_{1-x} materials with applied stress (**a**) and reciprocal temperature (**b**): $HfC_{0.91}$ (sintered and annealed in vacuum, porosity $\leq$ 10%, mean grain size – 22 μm, contents O + N – 0.24%) [144] and $HfC_{0.98}$ (sintered, porosity – 5%, mean grain size – 22 μm, contents: non-combined C – 0.05%, O – 0.10%, N – 0.01%, W – 0.02%) [328]

Table 3.13 Formal creep characteristics (activation energy Q, stress exponent constant n) of hafnium monocarbide phases HfC_{1-x} at various temperature and stress ranges[a]

Composition	Load type[b]	Temperature range, °C	Stress range, MPa	Activation energy[c], Q, kJ mol^{-1}	Exponent constant[c], n	References
$HfC^{d}_{0.96–0.99}$	C	2400–2700	8–70	830 ± 60	1	[142]
$HfC^{e}_{0.98}$	C	2200–2400	60–100	435	2.6	[328–330]
	C	2200–2400	100–200	810	6.5	[328–330]
	C	2600–2800	60–100	810	5–6.5	[328–330]
$HfC^{f}_{0.94}$	F	2500–3050	10–16	815	1	[145]
$HfC^{g}_{0.93}$	F	2500–2950	10–17.5	815 ± 40	1	[143]
$HfC^{h}_{0.91}$	C	1950–2100	50	840	–	[144]

[a]For near-stoichiometric HfC_{1-x} the following average values were estimated: Peierls stress τ_P = 6.0 GPa, shear modulus – Peierls stress ratio τ_P/G = 0.032 and slip activation energy barrier q = 1.2 eV [392]
[b]Denoted: F – flexure (bending), C – compression
[c]*See* Eq. (I-3.17)
[d]Sintered, porosity – 3%, mean grain size – 11–15 μm, contents: O/Hf and N/Hf ratios $\leq$0.03
[e]Sintered, porosity – 5%, mean grain size – 22 μm, contents: non-combined C – 0.05%, O – 0.10%, N – 0.01%, W – 0.02%
[f]Hot-pressed, porosity – 3–5%
[g]Hot-pressed, porosity $\leq$5%, mean grain size – 80 μm, contents: O $\leq$ 0.1%, N $\leq$ 0.09%
[h]Sintered and annealed, porosity $\leq$10%, mean grain size – 22 μm, O + N – 0.24%

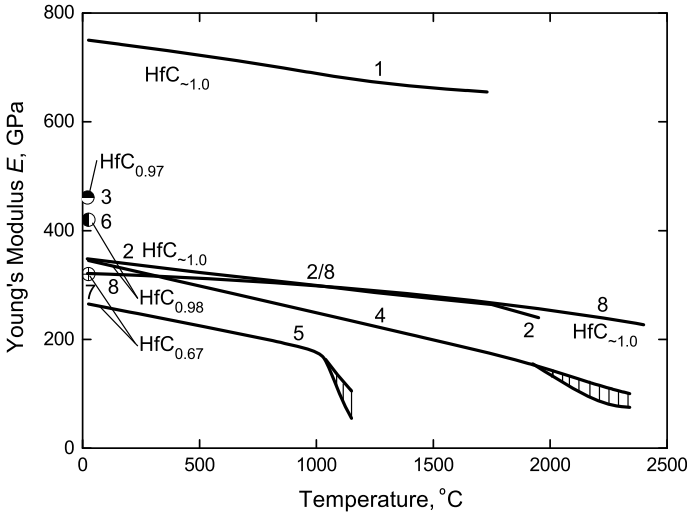

Fig. 3.20 Variations of Young's modulus E with temperature of hafnium carbide HfC_{1-x} phases: 1 – recommended by Kotelnikov et al. [31]; 2 – calculated on the basis of hardness-elasticity relationship by Travushkin et al. [126, 127]; 3 – hot-pressed and sintered materials, porosity – 4–7%, determined by the slope of stress-strain curves employing 4-point bending scheme by Sanders et al. [79]; 4–7 – hot-pressed materials, measured by Opeka et al. (4–5 – from flexural testing, upper lines correspond to the initial modulus (without plastic components) and lower lines correspond to the secondary modulus (with plastic components), 6 – from tensile testing of ring specimens and 7 – from ultrasonic measurements) [91]; 8 – hot-pressed materials, measured by Jun and Shaffer (resonant frequency technique) [82, 311]

The variations of Young's modulus E with temperature for the HfC_{1-x} phases with different deviations from the stoichiometry are shown in Fig. 3.20; the elastic properties (stiffness coefficients c_{ij} and moduli E, K and Poisson's ratio ν) of various HfC_{1-x} materials are listed in Table 3.14. The variations of the elastic characteristics with higher pressure (up to 100 GPa) for quasi-stoichiometric hafnium monocarbide $HfC_{\sim 1.0}$ based on *ab initio* calculations are shown in Fig. 3.21. According to the calculations on the basis of DFT with LDA produced by Krasnenko and Brik [443] bulk (compression) modulus K of $HfC_{\sim 1.0}$ arises at elevated hydrostatic pressures p (up to 50 GPa) linearly

$$K = 258.0 + 4.21p. \tag{3.38}$$

The testing data on thermal shock/stress resistance (thermal strength) of single phase hafnium monocarbide HfC_{1-x} materials are given in Table 3.15.

Table 3.14 Elastic properties (stiffness coefficients c_{11}, c_{12} and c_{44}, Young's modulus E, bulk (compression) modulus K, Poisson's ratio v)[a] of hafnium monocarbide phases HfC_{1-x}

Composition	Stiffness coefficients c_{ij}			Young's modulus E, GPa	Bulk modulus K, GPa	Poisson's ratio v	References
	c_{11}, GPa	c_{12}, GPa	c_{44}, GPa				
$HfC_{0.67}$	–	–	–	265[b,c]	–	–	[91]
	–	–	–	317[b,d]	–	–	[91]
	–	–	–	320[b,e]	–	–	[91]
$HfC^{f}_{0.75}$	464	101	131	366[g]	223[g]	–	[334]
$HfC^{f}_{0.88}$	508	107	162	425[h]	223[h]	–	[334]
$HfC^{i}_{0.96}$	–	–	–	430[j]	203[j]	–	[441]
$HfC_{0.97}$	535[f]	110[f]	170[f]	445[f,k]	252[f,k]	–	[334]
	–	114[l,m,n]	195[l,m]	461[l,m,o]	242[l,m,o]	0.18[l,m]	[3, 9, 82, 150, 314]
	–	–	–	430[p]	–	–	[151]
	–	–	–	321[q]	–	–	[79]
	–	–	199	470	246	0.18	[4]
	–	–	–	–	263	–	[157, 372]
$HfC_{0.98}$	–	–	–	345[c,r]	–	–	[91]
	–	–	–	420[d,r]	–	–	[91, 381]
	–	–	–	427[e,r]	–	–	[91]
$HfC_{\sim 1.0}$	–	–	179[s]	317–421[s]	291[s]	–	[9, 152]
	–	–	–	424[t]	–	0.17	[154]
	–	–	–	351[u]	–	–	[67, 108]
	–	–	–	480 ± 10[v]	–	–	[156]
	–	–	–	352	–	–	[35, 155]
	–	–	–	514[n]	–	–	[1]
	–	–	193	470	275	0.215	[4]
	–	–	–	422	–	–	[14, 15]
	500[w]	114[w]	180[w]	433	243[w]	–	[159, 392, 515]
	574[x]	107[x]	180[x]	–	260[x]	–	[161]
	–	–	–	–	238[x]	–	[162, 331]
	591[y]	119[y]	163[y]	461[y,z]	276[y,z]	0.22[y]	[39]
	527[a1]	107[a1]	160[a1]	431[a1,a2]	247[a1,a2]	0.21[a1]	[39]
	506[a3]	105[a3]	160[a3]	422[a3,a4]	238[a3,a4]	0.205[a3]	[39]
	528[a5]	111[a5]	160[a5]	431[a2,a5]	250[a2,a5]	0.21[a5]	[39]
	–	–	–	404[a6,a7]	236[a6,a7]	0.215[a6]	[62]
	577[a8]	117[a8]	171[a8]	537[a8,a9]	270[a8,a9]	–	[164]
	579[a5]	103[a5]	204[a5]	–	260[a5]	–	[287]
	–	–	–	–	257[y]	–	[288]
	–	–	–	–	238[b1]	–	[288]
	–	–	–	–	218[f]	–	[288]
	554[b2]	132[b2]	228[b2]	–	–	–	[295]
	514[b3]	98[b3]	178[b3]	449[b3,b4]	237[b3,b4]	0.18[b3]	[315]
	486[a8]	106[a8]	156[a8]	–	233[a8]	–	[316]
	480[a8]	103[a8]	177[a8]	–	–	–	[317]

(continued)

Table 3.14 (continued)

Composition	Stiffness coefficients c_{ij}			Young's modulus E, GPa	Bulk modulus K, GPa	Poisson's ratio ν	References
	c_{11}, GPa	c_{12}, GPa	c_{44}, GPa				
	570[a8,b5]	140[a8,b5]	214[a8,b5]	513[a8,b5,b6]	283[a8,b5,b6]	0.20[a8,b5]	[318, 319]
	428[a8,b7]	100[a8,b7]	149[a8,b7]	373[a8,b7,b8]	210[a8,b7,b8]	0.20[a8,b7]	[318, 319]
	533[a8]	104[a8]	162[a8]	437[a8,b9]	247[a8,b9]	0.205[a8]	[320]
	498[c1]	113[c1]	179[c1]	455[c1,c2]	241[c1,c2]	0.19[c1]	[163]
	548[a8]	99[a8]	192[a8]	518[a8,c3]	248[a8,c3]	0.15[a8]	[163]
	–	–	–	255[c4]	–	–	[321]
	454[c5]	87[c5]	87[c5]	426[c5,c6]	218[c5,c6]	0.16[c5]	[322]
	465[c1]	79[c1]	91[c1]	–	–	–	[323]
	556[y]	105[y]	152[y]	–	248[y]	–	[324]
	502[a5]	97[a5]	143[a5]	–	–	–	[324]
	–	–	–	–	265[c7]	–	[325]
	537[c8]	112[c8]	156[c8]	498[c8,c9]	254[c8,c9]	0.17[c8]	[305]
	–	–	–	–	243[d1]	–	[327]
	–	–	–	–	242[d2]	–	[327]
	–	–	–	–	310[f]	–	[38]
	–	–	–	–	314[d3]	–	[38]
	540[f]	112[f]	171[f]	–	253[f,d4]	–	[334]
	609[f]	112[f]	183[f]	–	278[f,d5]	–	[333]
	582[d6]	103[d6]	173[d6]	–	263[d6]	–	[336, 443]
	507[f]	103[f]	165[f]	–	238[f]	–	[336, 443]
	–	–	–	–	242–275[d7]	–	[336]
	–	–	–	355	–	–	[386]
	–	–	–	464	–	–	[387]
	–	–	–	471	–	–	[388]
	474[d8]	88[d8]	98[d8]	337[d8,d9]	216[d8,d9]	0.23[d8]	[391, 392]
	–	–	–	450[e1]	–	–	[104]
	–	–	–	386[e2]	–	–	[439]
	–	–	–	–	246[e3]	–	[440]
	551[e4]	105[e4]	175[e4]	461[e4,e5]	254[e4,e5]	0.197[e4]	[442]
	537[f]	112[f]	156[f]	498[f,c9]	254[f,c9]	0.173[f]	[444]
	–	–	–	350[e6]	–	–	[153, 492]
	–	–	–	283 ± 10[e7]	–	–	[539]
	–	–	–	–	242[f,a2]	–	[326]
	–	–	–	740–750 (?)	–	–	[31, 158]

[a]For isotropic (or quasi-isotropic) materials [3]: $E = 2G(1 + \nu)$, $E = 3K(1 - 2\nu)$, $K = c_{12} + 2c_{44}/3$, $G = c_{44}$; the condition for the isotropy is given in Eq. (I-2.18)
[b]Hot-pressed, mean grain size >200 μm, porosity – 5–10%
[c]From flexural stress versus deflection measurements
[d]From ultrasonic measurements
[e]From ring specimens tensile testing measurements
[f]Calculated on the basis of density-functional theory (DFT) with generalized gradient approximation (GGA) using Perdew-Burke-Ernzerhof (PBE) functional
[g]Calculated value of G = 149 GPa
[h]Calculated value of G = 176 GPa
[i]Hot-pressed, corrected to the poreless state (true porosity – up to 12%), mean grain size – 5–20 μm; ultrasonic wave velocity measurements
[j]Measured experimentally G = 193 GPa

(continued)

Table 3.14 (continued)

[k]Calculated value of G = 184 GPa
[l]Hot-pressed, porosity – 3%, corrected to the poreless state
[m]From ultrasonic measurements, corrected to the poreless state
[n]Calculated value
[o]Measured experimentally G = 195 GPa
[p]Hot isostatically pressed and subsequently sintered, non-combined C – 0.05%
[q]Hot-pressed, mean grain size – 13 μm, porosity – 4–7%, determined by the slope of stress-strain curves employing 4-point bending scheme
[r]Hot-pressed, mean grain size – 40–60 μm, porosity – 5–10%
[s]Self-bonded sintered
[t]Porosity – 5.0%
[u]Hot-pressed, porosity – 6–8%
[v]Sintered in vacuum and annealed, mean grain size – 3.0 μm, porosity – 5%, contents: non-combined C – 0.09%, O – 0.09%, N – 0.05%
[w]Estimation from phonon dispersion curves (c_{11} is extrapolated from neutron data [160])
[x]Calculated on the basis of density-functional perturbation theory (DFPT)
[y]Calculated on the basis of DFT with local density approximation (LDA)
[z]Calculated value of G = 189 GPa
[a1]Calculated on the basis of DFT with LDA together with PBE functional
[a2]Calculated value of G = 178 GPa
[a3]Calculated on the basis of DFT with LDA together with the revised Perdew-Burke-Ernzerhof (RPBE) functional
[a4]Calculated value of G = 175 GPa
[a5]Calculated on the basis of DFT using plane-wave pseudopotential (PWPP) method with GGA
[a6]Calculated using the Debye–Grüneisen model combined with *ab initio* calculations
[a7]Calculated value of G = 166 GPa
[a8]From *ab initio* calculations
[a9]Calculated value of G = 230 GPa
[b1]Calculated on the basis of DFT with Wu-Cohen GGA exchange-energy functional
[b2]At 80 K
[b3]Calculated on the basis of full-potential *ab initio* approach
[b4]Calculated value of G = 190 GPa
[b5]At 0 K
[b6]Calculated value of G = 215 GPa
[b7]At ~300 K
[b8]Calculated value of G = 155 GPa
[b9] Calculated value of G = 181 GPa
[c1]Calculated on the basis of three body force potential (TBFP) model
[c2]Calculated value of G = 184 GPa and $\kappa = 4.15\ \mathrm{TPa}^{-1}$
[c3]Calculated value of G = 205 GPa and $\kappa = 4.03\ \mathrm{TPa}^{-1}$
[c4]Coatings synthesized by C_2H_2-reactive magnetron sputtering
[c5]Calculated on the basis of interionic potential theory with modified ionic charge
[c6]Calculated value of G = 125 GPa
[c7]First principles calculations using the plane-wave pseudopotential method
[c8]First principles calculations
[c9]Calculated value of G = 177 GPa
[d1]Calculated on the basis of DFT with GGA together with the Perdew's functional
[d2]Calculated on the basis of DFT with GGA using RPBE functional
[d3]Calculated on the basis of DFT with LDA using Ceperley-Alder (CA) type parametrization
[d4]Calculated value of G = 187 GPa
[d5]Calculated value of G = 207 GPa
[d6]On the basis of DFT with LDA using Ceperley-Alder-Perdew-Zunger (CA-PZ) functional
[d7]Calculated on the basis of Murnaghan equation of state

(continued)

Table 3.14 (continued)

[d8]Calculated on the basis of a modified interaction potential model with covalency ($MIPMC_v$)
[d9]Calculated value of G = 136 GPa
[e1]Calculated by the data acquisition software of the nanoindentation measurements based on Oliver-Pharr model [437]
[e2]Hot-pressed, porosity – 6%; calculated from depth-sensing indentation (DSI) measurements
[e3]On the basis of interconsistency of the physical properties of transition metal monocarbides
[e4]Calculated on the basis of DFT by projector augmented wave (PAW) method with GGA using PBE functional
[e5]Calculated value of G = 192 GPa
[e6]Hot-pressed; experimentally measured by the method of continuous impression of an indenter
[e7]Spark-plasma sintered, mean grain size – 1.8 ± 0.2 μm, porosity – 15%; measured experimentally by Berkovich nanoindentation

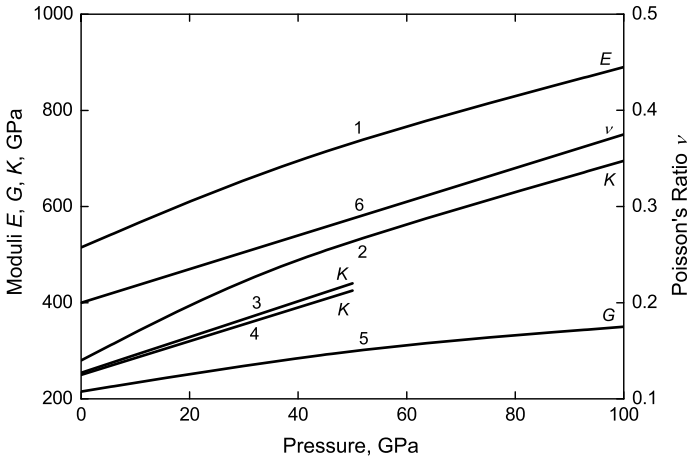

Fig. 3.21 Variations of elastic properties (1 – Young's modulus E, 2–4 – bulk (compression) modulus K, 5 – Coulomb's (shear) modulus G and 6 – Poisson's ratio ν) of quasi-stoichiometric hafnium monocarbide $HfC_{\sim 1.0}$ with higher pressure (at 0 K) based on *ab initio* calculations: 1–2, 5–6 – by Varshney et al. [318, 319]; 3 – by Gautam and Kumar [442] and 4 – by Krasnenko and Brik [336, 443]

Table 3.15 Thermal shock/stress resistance (thermal strength) testing data on single phase hafnium monocarbide HfC_{1-x} materials

Materials and its characterization	Specimen shape and dimensions, mm	Method of thermal loading (testing)	Experimental results	References
$HfC_{\sim 1.0}$, gas-phase deposited, contents: O + N ≤ 0.3%	Hollow cylinders: wall thickness – 0.8–1.2, height – 20	Direct resistance heating with 400 K s^{-1}, cooling by inert gas blow, cycle time – 3 min	Disintegration after 2–3 heating-cooling cycles from 2870 to 50 °C	[4, 141]

The magnitudes of physico-mechanical (strength, elasticity) properties of near-stoichiometric hafnium carbide HfC_{1-x} in the wide range of temperatures are summarized in Addendum in comparison with other ultra-high temperature materials.

3.5 Nuclear Physical Properties

The isotopes of element hafnium Hf (standard atomic mass – 178.49 u) from ^{153}Hf to ^{188}Hf, including metastable isomers (^{153m}Hf, ^{156m}Hf, ^{171m}Hf, ^{172m}Hf, $^{174m1-4}$Hf, $^{177m1-3}$Hf, $^{178m1-3}$Hf, $^{179m1-2}$Hf, $^{180m1-5}$Hf, $^{181m1-3}$Hf, ^{182m}Hf and ^{184m}Hf), and their general characteristics are summarized in Table 3.16; neutron nuclear physical properties of the isotopes of hafnium are given in Table 3.17. The comprehensive list of isotopes of element carbon C is presented in Table I-2.12 and its nuclear physical properties, including isotopic mass range, total number of isotopes, thermal neutron macroscopic cross sections, moderating ability and capture resonance integral, are given in Table I-A.8.

The thermal neutron macroscopic cross sections Σ_i (*see* Eq. 2.53) of near-stoichiometric hafnium monocarbide HfC_{1-x} (for 2200 m s^{-1} neutrons) [31]:

cross section of capture (absorption) Σ_a, cm^{-1}	4.17
cross section of scattering Σ_s, cm^{-1}	~0.59

For the estimation of probable damage of hafnium monocarbide materials exposed to various types of radiation the parameters of formation and migration of lattice point defects (vacancies and interstitial atoms) are given in Table 3.18.

Table 3.16 General characteristics of the isotopes of hafnium [165–169]

Isotope	Mass, u	Abundance, %	Half-life period	Decay mode, excitation (radiation) energy, MeV
^{180}Hf[a]	179.946550	35.08	–	–
^{178}Hf[a]	177.943699	27.28	–	–
^{177}Hf[a]	176.943221	18.60	–	–
^{179}Hf[a]	178.945816	13.62	–	–
^{176}Hf[a]	175.941409	5.26	–	–
^{174}Hf[b]	173.940046	0.16	2.0×10^{15} year	α
^{182}Hf[c]	181.950554	–	8.9×10^{6} year	β^-
^{178m2}Hf	–	–	31 year	Isomer, γ, 2,446
^{172}Hf	171.939448	–	1.87 year	*K*-capture; γ, 0.28, 0.8
^{175}Hf	174.941509	–	70 day	β^+
^{181}Hf	180.949101	–	42.39 day	β^-, 0.408; γ, 0.004, 0.133, 0.136, 0.137, 0.346, 0.473, 0.482, 0.615
^{179m2}Hf	–	–	25.05 day	Isomer, γ, 1.106
^{173}Hf	172.94051	–	23.6 h	β^+
^{170}Hf	169.93961	–	16.01 h	*K*-capture
^{171}Hf	170.94049	–	12.1 h	β^+
^{180m1}Hf	–	–	5.47 h	Isomer, γ, 1.141
^{184}Hf	183.95545	–	4.12 h	β^-
^{183}Hf	182.95353	–	1.067 h	β^-
^{182m}Hf	–	–	1.025 min	β^- (58%); Isomer (42%), 1.172
^{177m3}Hf	–	–	51.4 min	Isomer, γ, 2.740

(continued)

Table 3.16 (continued)

Isotope	Mass, u	Abundance, %	Half-life period	Decay mode, excitation (radiation) energy, MeV
^{168}Hf	167.94057	–	25.95 min	β^+
^{166}Hf	165.94218	–	6.77 min	β^+
^{185}Hf	184.95882 (?)	–	3.5 min	β^-
^{169}Hf	168.94126	–	3.24 min	β^+
^{186}Hf	185.96089 (?)	–	2.6 min	β^-
^{167}Hf	166.94260	–	2.05 min	β^+
^{164}Hf	163.944367	–	111 s	β^+
^{165}Hf	164.94457	–	76 s	β^+
^{184m}Hf	–	–	48 s	Isomer, β^-, 1.272
^{163}Hf	162.94709	–	40.0 s	β^+, α (10^{-4}%)
^{162}Hf	161.94721	–	39.4 s	β^+ (99.99%), α (0.008%)
^{187}Hf	186.96459 (?)	–	30 (?) s	?
^{171m}Hf	–	–	29.5 s	Isomer, γ, 0.022
^{188}Hf	187.96685 (?)	–	20 (?) s	?
^{179m1}Hf	–	–	18.67 s	Isomer, γ, 0.375
^{161}Hf	160.950275	–	18.2 s	β^+ (99.7%), α (0.3%)
^{160}Hf	159.950684	–	13.6 s	β^+ (99.3%), α (0.7%)
^{159}Hf	158.953995	–	5.20 s	β^+ (59%), α (41%)
^{178m1}Hf	–	–	4.0 s	Isomer, γ, 1.147
^{158}Hf	157.954799	–	2.84 s	β^+ (55%), α (45%)
^{154}Hf	153.96486 (?)	–	2 s	β^+, α (rare)
^{177m1}Hf	–	–	1.09 s	Isomer, γ, 1.315
^{155}Hf	154.96339 (?)		890 ms	β^+, α (rare)
^{153m}Hf	–	–	500 (?) ms	Isomer, γ, 0.750 (?)
^{153}Hf	152.97069 (?)	–	400 (?) ms	?
^{157}Hf	156.95840 (?)	–	115 ms	α (86%), β^+ (14%)
^{156}Hf	155.95936	–	23 ms	α (97%), β^+ (3%)
^{181m3}Hf	–	–	1.5 ms	Isomer, γ, 1.738
^{156m}Hf	–	–	480 μs	Isomer, γ, 1.959
^{181m2}Hf	–	–	~100 μs	Isomer, γ, 1.040
^{180m6}Hf	–	–	90 μs	Isomer, γ, 3.599
^{181m1}Hf	–	–	80 μs	Isomer, γ, 0.595
^{178m3}Hf	–	–	68 μs	Isomer, γ, 2.574
^{177m2}Hf	–	–	55.9 μs	Isomer, γ, 1.342
^{180m3}Hf	–	–	15 μs	Isomer, γ, 2.426
^{180m5}Hf	–	–	>10 μs	Isomer, γ, 2.538
^{180m4}Hf	–	–	10 μs	Isomer, γ, 2.486
^{174m4}Hf	–	–	3.7 μs	Isomer, γ, 3.312
^{174m2}Hf	–	–	2.39 μs	Isomer, γ, 1.798
^{174m3}Hf	–	–	2.39 μs	Isomer, γ, 1.798
^{180m2}Hf	–	–	0.57 μs	Isomer, γ, 1.374
^{172m}Hf	–	–	163 ns	Isomer, γ, 2.006
^{174m1}Hf	–	–	138 ns	Isomer, γ, 1.549

[a]Believed to undergo α decay (?)
[b]Primordial radionuclide
[c]Believed to occur naturally after the α decay of ^{186}W (?)

Table 3.17 Neutron nuclear physical properties of the isotopes of hafnium [170, 171, 445, 446, 620]

Isotope	Microscopic thermal neutron cross sections[a], b		Macroscopic thermal neutron cross sections[a], cm^{-1}		Resonance integral for neutron capture I_γ, b
	Capture (absorption) σ_a	Scattering[b] σ_s	Capture (absorption) Σ_a	Scattering[b] Σ_s	
Hf[c]	104.1 ± 0.5	10.2 ± 0.4	4.67	0.45	1990 ± 50
^{180}Hf	13.04 ± 0.07	22 ± 1	–	–	35 ± 1
^{178}Hf	84 ± 4	4.4 ± 0.3	–	–	1950 ± 120
^{177}Hf	370 ± 10	0.2 ± 0.2	–	–	7200 ± 200
^{179}Hf	41 ± 3	7.1 ± 0.3	–	–	630 ± 30
^{176}Hf	24 ± 3	5.5 ± 0.3	–	–	880 ± 40
^{174}Hf	549 ± 7	15 ± 3	–	–	440 ± 40

[a]For 2200 m s^{-1} neutrons
[b]Total bound scattering cross sections
[c]Occurred naturally

Table 3.18 Parameters of formation and migration of lattice point defects (vacancies and interstitial atoms) in near-stoichiometric hafnium monocarbide[a]

Defect	Metal sublattice				Non-metal sublattice			
	E_f, eV	E_m, eV	S_f/k_B	S_m/k_B	E_f, eV	E_m, eV	S_f/k_B	S_m/k_B
Vacancy[b] [172, 546]	3.98	3.32	4.38	3.65	1.33	5.32	1.46	5.84
Vacancy[c] [373, 546]	3.67	3.10	4.03	3.41	1.04	3.94	1.14	4.33
Vacancy[d] [334]	9.3	–	–	–	1.1	4.8[e]	–	–
Vacancy[f] [545]	4.24	–	–	–	–	–	–	–
Di-vacancy[g] [621]	7.30	–	–	–	7.30	–	–	–
Interstitial atom[b] [172]	19.29	0.66	21.20	0.73	5.99	0.66	6.58	0.73

[a]Denoted: E_f – defect formation energy, E_m – defect migration energy, S_f – defect formation entropy, S_m – defect migration entropy, k_B – Boltzmann constant
[b]Calculated on the basis of bonding model (relaxation displacement of atoms surrounding a defect is not taking into account)
[c]Calculated on the basis of elastic continuum model
[d]*Ab initio* simulation computed from the largest super-cell on the basis of the PAW pseudopotential using the PBE formulation
[e]Calculated value of activation energy E_A = 570 kJ mol^{-1} (*see* also Table 3.23)
[f]Calculated on the basis of data on the yield strength determined by hardness and elasticity experimental measurements
[g]Calculated on the basis of PAW within GGA

Nuclear physical properties of hafnium monocarbide HfC_{1-x} in comparison with other ultra-high temperature materials are also given in Addendum.

3.6 Chemical Properties and Materials Design

The information on the chemical properties, compatibility and interaction of hafnium monocarbide HfC_{1-x} at elevated, high and ultra-high temperatures with elements (metals, non-metals) is summarized in Table 3.19, with refractory

Table 3.19 Chemical interaction and/or compatibility of hafnium carbide HfC_{1-x} with elements (metals, non-metals) at elevated, high and ultra-high temperatures, including solid matters and molten media (reaction systems are given mainly in alphabetical order)[a]

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	References
HfC_{1-x}–Al	–	<1000	No chemical interaction in the contact zone; lattice parameter of HfC_{1-x} is invariable	[1, 31, 173, 368, 447–449, 526, 549, 564]
	–	>1250	The formation of solid solutions (?); lattice parameter of HfC_{1-x} is decreasing slightly	
	–	1400–2800	Addition of 1% Al does not affect the densification of powdered carbide by hot-pressing	
	–	1700	Testing in Al vapours results in the formation of refractory phase impregnated with Al in pores	
	–	–	Formation of $HfAl_3C_3$, $HfAl_4C_4$, $Hf_2Al_3C_4$, $Hf_2Al_4C_5$, $Hf_3Al_3C_5$, $Hf_3Al_4C_6$ and Hf_5Al_3C *See also* section HfC_{1-x}–Al_4C_3 in Table 3.20 *See also* Table 3.22 *See also* section C–Al–Hf in Table I-2.14	
HfC_{1-x}–SiC–Al	–	–	Formation of $Hf(Al_{1-x}Si_x)_4C_4$, $Hf_2(Al_{1-x}Si_x)_4C_5$ and $Hf_3(Al_{1-x}Si_x)_4C_6$ *See also* section HfC_{1-x}–Al_4C_3–Si in this table *See also* section HfC_{1-x}–Al_4C_3–SiC in Table 3.20	[526]
HfC_{1-x}–SiC–TiC_{1-x}–Al	–	–	Formation of solid solutions $(Ti_{1-x}Hf_x)_3(Si_{1-x}Al_x)C_2$ based on Ti_3SiC_2 *See also* section HfC_{1-x}–Al_4C_3–SiC–TiC_{1-x} in Table 3.20	[514]
HfC_{1-x}–β-B	–	1400–1500	The max. solubility of B in HfC_{1-x} is ~6 mol.% (?)	[31, 194, 527, 571]
	–	1500	Intensive interaction (small mass loss) with the formation of $HfB_{2\pm x}$ *See also* section C–B–Hf in Table I-2.14	
HfC_{1-x}–Be	–	–	The effect of Be as an additive for hot-pressing process was studied	[31]
HfC_{1-x}–Bi			*See* Table 3.22	
HfC_{1-x}–C	–	~3100–3280	Eutectic HfC_{1-x}–C (graphite)	[4, 12, 17, 22, 201, 343, 344, 518, 536]
			See also section C–Hf in Table I-2.13	
HfC_{1-x}–C–Fe			*See* Table 3.22	
HfC_{1-x}–C–Fe–Mn			*See* Table 3.22	

(continued)

Table 3.19 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	References
HfC_{1-x}–Cd	–	450	No chemical interaction	[368]
HfC_{1-x}–α/ε-Co	–	~1370–1420	Eutectic HfC_{1-x}–α-Co	[1, 13, 31, 35, 204, 455, 532]
	–	1500	The contact zone between bulk HfC_{1-x} and metallic Co includes eutectic alloy	
	–	–	Addition of 1% Co intensifies the densification of powdered carbide by hot-pressing	
			See also Table 3.22	
			See also section C–Co–Hf in Table I-2.14	
HfC_{1-x}–NbC_{1-x}–Co	–	~1360–1435	Eutectic $(Hf,Nb)C_{1-x}$ (x = 0.06–0.18) – α-Co	[174]
			See also section C–Co–Hf–Nb in Table I-2.14	
HfC_{1-x}–TaC_{1-x}–Co			*See* section TaC_{1-x}–HfC_{1-x}–Co in Table 2.21	
			See also section C–Co–Hf–Ta in Table I-2.14	
HfC_{1-x}–TaC_{1-x}–TiC_{1-x}–δ-$TiN_{1±x}$–δ-$WC_{1±x}$–β-$Mo_{2±x}C$–Co			*See* section TaC_{1-x}–HfC_{1-x}–TiC_{1-x}–δ-$TiN_{1±x}$–δ-$WC_{1±x}$–β-$Mo_{2±x}C$–Co in Table 2.21	
HfC_{1-x}–TiC_{1-x}–Co	–	1370	Eutectic $(Hf,Ti)C_{1-x}$ (x = 0.06)–$(Ti,Hf)C_{1-x}$ (x = 0.20)–α-Co	[175]
			See also section C–Co–Hf–Ti in Table I-2.14	
HfC_{1-x}–VC_{1-x}–Co	–	1320	Eutectic $(Hf,V)C_{1-x}$ (x = 0.06)–$(V,Hf)C_{1-x}$ (x = 0.17)–α-Co	[175]
			See also section C–Co–Hf–V in Table I-2.14	
HfC_{1-x}–ZrC_{1-x}–Co	–	~1380–1435	Eutectic $(Hf,Zr)C_{1-x}$ (x = 0.06–0.19)–α-Co	[174]
			See also section C–Co–Hf–Zr in Table I-2.14	
HfC_{1-x}–Cr	Ar	1790–1830	Eutectic HfC_{1-x}–Cr; the max. solubility of Cr in HfC_{1-x} is ~9 mol.% and that of HfC_{1-x} in Cr is ~1 mol.%	[1, 13, 176, 204, 504, 552]
	–	–	Stable and compatible with each other as cermet components	
			See also section C–Cr–Hf in Table I-2.14	
HfC_{1-x}–Cs	Cs vapour, ~13 Pa	~2000	No chemical reaction observed, no effect upon structural characteristics of HfC_{1-x} and its thermionic emission parameters (some tens of hours exposure)	[119, 339]
HfC_{1-x}–Cu			*See* Table 3.22	
HfC_{1-x}–Cu–Hf			*See* Table 3.22	
HfC_{1-x}–Cu–Hf–V			*See* Table 3.22	

(continued)

Table 3.19 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	References
HfC_{1-x}–α/γ/δ-Fe	–	~1410–1490	Eutectic HfC_{1-x}–δ-Fe; the max. solubility of HfC_{1-x} in δ-Fe is ≤ 0.3 mol.%	[1, 13, 31, 35, 204, 205, 464, 532, 554]
	–	1550	The contact zone between bulk HfC_{1-x} and metallic Fe includes eutectic alloy	
	–	–	DFT calculated semicoherent interfacial energy at relaxed interface – 0.95 J m^{-2}	
			See also Table 3.22	
			See also section C–Fe–Hf in Table I-2.14	
HfC_{1-x}–Fe–Mn			*See* Table 3.22	
HfC_{1-x}–Ga			*See* Table 3.22	
HfC_{1-x}–Ge			*See* Table 3.22	
HfC_{1-x}–α/β-Hf	–	~2240–2360	Peritectic α-Hf (in equilibrium with $HfC_{0.52}$)	[8–11, 31, 204]
			See also section C–Hf in Table I-2.13	
HfC_{1-x}–In	–	–	Formation of Hf_2InC ($M_{n+1}AX_n$-phase)	[177, 178, 628]
			See also Table 3.22	
			See also section C–Hf–In in Table I-2.14	
HfC_{1-x}–TiC_{1-x}–In	–	–	Formation of $(Hf,Ti)_2InC_{1+x}$ ($M_{n+1}AX_n$-phase solid solution); $(Hf_{0.5}Ti_{0.5})InC$ and $(Hf_{0.53}Ti_{0.47})InC_{1.26}$ were synthesized	[177, 178, 529–531]
			See also section C–Hf–In–Ti in Table I-2.14	
HfC_{1-x}–Ir	–	1650–1930	Near-stoichiometric HfC_{1-x} reacts with Ir to form $HfIr_{3+x}$ ($x \approx 0$) containing C deposits of varying morphology: large widely spaced at lower temperatures and smaller in size uniformly dispersed at higher temperatures	[13, 202]
			See also section C–Hf–In in Table I-2.14	
HfC_{1-x}–K	–	~2000	No chemical reaction observed, no effect upon structural characteristics of HfC_{1-x} and its thermionic emission parameters (some tens of hours exposure)	[119, 339]
HfC_{1-x}–Li	–	–	HfC_{1-x} is highly resistant to Li liquid environments	[393, 394]
HfC_{1-x}–α/β/γ/δ-Mn	–	–	Intensive interaction; the formation of eutectic alloy and solid solutions	[1, 31, 341]
	–	–	Formation of metastable Mn_3Hf_3C (η-carbide type, ?)	
	–	–	The effect of Mn as an additive for hot-pressing process was studied	
			See also Table 3.22	
			See also section C–Hf–Mn in Table I-2.14	

(continued)

Table 3.19 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	References
HfC_{1-x}–Mo	–	1250	The max. solubility of HfC_{1-x} in Mo is 0.25 mol.% (?)	[1, 13, 31, 53, 179–182, 186, 195, 204, 342, 552, 569, 570]
	–	1500–2200	Formation of solid solutions (exposure 5 h)	
	Vacuum	≥1800	The initiation of reaction between powdered HfC_{1-x} and compact dense Mo (exposure 5 h)	
	–	2000–2200	Noticeable solid state interaction between bulk and dense materials	
	–	2100	The max. solubility of HfC_{1-x} in Mo is 0.4 mol.%	
	–	–	Stable and compatible with each other as cermet components	
	–	~2310–2370	Eutectic HfC_{1-x}–Mo; the max. solubility of Mo in HfC_{1-x} is ≥15 mol.% and that of HfC_{1-x} in Mo is 1.6 mol.% *See also* Table 3.22 *See also* section C–Hf–Mo in Table I-2.14	
HfC_{1-x}–Mo–Ni	Vacuum, 0.01–0.1 Pa	1450	The contact interaction of compact dense HfC_{1-x} (x = 0.05) with Ni–25% Mo melt leads to the preferential dissolution of C; after cooling the isolations based on β-$Mo_{2\pm x}C$ (hexagonal) phase, surrounded by metal-semicarbide eutectics, are formed in the metallic melt based on Ni (Hf content ≤2%) in parallel with Ni–$HfNi_5$ eutectics	[624]
HfC_{1-x}–TiC_{1-x}–Mo–Ni	Vacuum, 0.01–0.1 Pa	1450	The interaction of $Hf_{0.21}Ti_{0.79}C_{0.95}$ phase with Ni–25% Mo melt is characterized by preferential dissolution of C and subsequent precipitation of $Hf_{0.01}Ti_{\sim 0.50}Mo_{\sim 0.50}C_{1-x}$ (from the melt saturated with Hf, Ti and C); after cooling the solidified Ni–$HfNi_5$ eutectics was revealed	[624]
HfC_{1-x}–TiC_{1-x}–δ-$TiN_{1\pm x}$–Mo–Ni	Vacuum, 0.01–0.1 Pa	1450	The interaction of $Hf_{0.20}Ti_{0.80}C_{0.75}N_{0.24}$ phase with molten Ni–20% Mo alloy leads to the accelerated incongruent dissolution of the components, preferentially – C and Hf, up to $Hf_{0.07}Ti_{0.93}C_{0.20}N_{0.80}$; the metallic melts have a hypereutectic structure with primary precipitates of $Hf_{0.02}Ti_{0.50}Mo_{0.48}$ $(C,N)_{1-x}$ and $Hf_{0.68}Ti_{0.14}Mo_{0.18}(C,N)_{1-x}$ in eutectics matrix	[623]
HfC_{1-x}–Nb	Vacuum	1000–1400	No interaction between powdered HfC_{1-x} and compact dense Nb	[1, 13, 31, 53, 195, 204]
	Vacuum	1400–2200	The formation of solid solutions	
	Vacuum	≥1400	The initiation of reaction between powdered HfC_{1-x} and compact dense Nb with the formation of solid solutions $(Hf,Nb)C_{1-x}$ and non-combined C (exposure 5 h) *See also* section C–Hf–Nb in Table I-2.14	

(continued)

Table 3.19 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	References
HfC_{1-x}–Ni	Ar	1180–1215	Formation of a liquid phase in the powdered mixture (mean grain size – ~50 μm) of the components	[1, 13, 31, 35, 183–185, 204, 450, 458–460, 532]
	–	~1310–1360	Eutectic HfC_{1-x}–Ni; the max. solubility of HfC_{1-x} in Ni is ≤0.7–1.0 mol.% and that of Ni in HfC_{1-x} is ~7.5 mol.%	
	–	1450	The contact zone between HfC_{1-x} and Ni includes eutectic alloy; carbide solid solutions are formed *See also* Table 3.22 *See also* section C–Hf–Ni in Table I-2.14	
HfC_{1-x}–β-$Mo_{2\pm x}C$–Ni	–	~1200–1240	Eutectic $(Hf,Mo)C_{1-x}$–β-$(Mo,Hf)_{2\pm x}C$–Ni *See also* section C–Hf–Mo–Ni in Table I-2.14	[218]
HfC_{1-x}–TaC_{1-x}–TiC_{1-x}–δ-$TiN_{1\pm x}$–Ni			*See* section TaC_{1-x}–HfC_{1-x}–TiC_{1-x}–δ-$TiN_{1\pm x}$–Ni in Table 2.21	
HfC_{1-x}–TaC_{1-x}–TiC_{1-x}–δ-$TiN_{1\pm x}$–δ-$WC_{1\pm x}$–β-$Mo_{2\pm x}C$–Ni			*See* section TaC_{1-x}–HfC_{1-x}–TiC_{1-x}–δ-$TiN_{1\pm x}$–δ-$WC_{1\pm x}$–β-$Mo_{2\pm x}C$–Ni in Table 2.21	
HfC_{1-x}–TiC_{1-x}–Ni	–	~1205–1235	Eutectic $(Hf,Ti)C_{1-x}$–$(Ti,Hf)C_{1-x}$–Ni (cast alloys show a tendency to crystallization in metastable state when the carbide phase does not decompose into two solid solutions)	[183, 451, 458–460, 624]
	Vacuum, 0.01–0.1 Pa	1450	$Hf_{0.20}Ti_{0.80}C_{0.95}$ phase has higher dissolution rate in molten Ni; preferentially C and Hf are transfered into the melt, when attaining composition $Hf_{0.05}Ti_{0.95}C_{<0.95}$ the dissolution becomes congruent *See also* section C–Hf–Ni–Ti in Table I-2.14	
HfC_{1-x}–TiC_{1-x}–Ni–Mo	Vacuum, 0.05 Pa	1450	The interaction of $Hf_{0.20}Ti_{0.80}C_{0.96}$ phase with Ni–20% Mo melt is characterized by preferential dissolution of C and subsequent precipitation of $Hf_{0.01}Ti_{\sim 0.60}Mo_{\sim 0.40}C_{1-x}$ (from the melt saturated with Hf, Ti and C)	[461, 624]
HfC_{1-x}–TiC_{1-x}–δ-$TiN_{1\pm x}$–Ni	Vacuum, 0.1–0.01 Pa	1450	The interaction of $Hf_{0.20}Ti_{0.80}C_{0.75}N_{0.24}$ phase with molten Ni leads to the incongruent dissolution of the components, preferentially – C and Hf, and simultaneous decomposition of the initial phase with the formation of isolated $Hf_{0.10}Ti_{0.90}(C,N)_{1-x}$ and $Hf_{0.55}Ti_{0.45}(C,N)_{1-x}$; the metallic melts have a hypoeutectic structure with primary precipitates of Ni based phase and ternary eutectics (one of its constituents is $Hf_{0.75}Ti_{0.25}(C,N)_{1-x}$)	[189, 227, 510, 553, 623]
	Vacuum	1450–1500	The interaction of $(Hf,Ti)C_{0.60–0.70}N_{0.30–0.40}$ carbonitride phases with Ni is considered	

(continued)

Table 3.19 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	References
HfC_{1-x}–TiC_{1-x}–δ-$TiN_{1\pm x}$–δ-$WC_{1\pm x}$–Ni	Vacuum	1450–1500	The interaction of (Hf,Ti,W)$C_{0.70}N_{0.30}$ carbonitride phase with Ni is considered	[510, 517]
HfC_{1-x}–δ-$WC_{1\pm x}$–Ni	–	~1000	Eutectic (Hf,W)C_{1-x}–δ-(W,Hf)$C_{1\pm x}$–Ni	[184]
			See also section C–Hf–Ni–W in Table I-2.14	
HfC_{1-x}–ZrC_{1-x}–Ni	Vacuum	1000	The max. solid solubility of (Zr,Hf)C_{1-x} in Ni is $\leq$1 mol.%	[185, 573]
	–	~1130–1330	Eutectic (Zr,Hf)C_{1-x} (x = 0.02–0.17)–Ni; the minimum on the line of eutectic solidification is 100–150 °C lower than the eutectic temperatures in ZrC_{1-x}–Ni and HfC_{1-x}–Ni systems	
			See also section C–Hf–Ni–Zr in Table I-2.14	
HfC_{1-x}–Os	–	–	The mutual solubilities of the components are low	[13]
			See also section C–Hf–Os in Table I-2.14	
HfC_{1-x}–Pb	–	450	No chemical interaction	[177, 178, 368]
	–	–	Formation of Hf_2PbC ($M_{n+1}AX_n$-phase)	
			See also Table 3.22	
			See also section C–Hf–Pb in Table I-2.14	
HfC_{1-x}–Pd			*See* section C–Hf–Pd in Table I-2.14	
HfC_{1-x}–Pt			*See* section C–Hf–Pt in Table I-2.14	
HfC_{1-x}–Re	–	2000–2200	The max. solubility of Re in HfC_{1-x} is 2 mol.%	[1, 13, 186, 199, 200, 204, 346, 552]
	He	2500	No interaction in the contact zone between the compact dense materials (exposure 1 h)	
		2600	The max. solubility of Re in HfC_{1-x} is 4 mol.%	
	–	–	Stable and compatible with each other as cermet components	
	–	2690–2750	Eutectic HfC_{1-x}–Re; the max. solid solubility of Re in HfC_{1-x} is ~4–5 mol.% and that of HfC_{1-x} in Re is also low	
			See also section C–Hf–Re in Table I-2.14	
HfC_{1-x}–Rh	–	–	No solubility of Rh in HfC_{1-x}	[13]
			See also section C–Hf–Rh in Table I-2.14	
HfC_{1-x}–Ru	–	~2000	Eutectic HfC_{1-x}–Ru	[13]
			See also section C–Hf–Ru in Table I-2.14	
HfC_{1-x}–S	–	–	Formation of Hf_2SC ($M_{n+1}AX_n$-phase)	[177, 178]
			See also section C–Hf–S in Table I-2.14	
HfC_{1-x}–Sb			*See* Table 3.22	
HfC_{1-x}–Si	–	$\geq$1500	Intensive interaction; formation of SiC and Hf silicides (or ternary phases ?)	[1, 12, 31, 187, 188, 368]
			See also Table 3.22	
			See also section C–Hf–Si in Table I-2.14	

(continued)

Table 3.19 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	References
HfC_{1-x}–Al_4C_3–Si	–	–	Formation of $Hf(Al_{1-x}Si_x)_4C_4$, $Hf_2(Al_{1-x}Si_x)_4C_5$ and $Hf_3(Al_{1-x}Si_x)_4C_6$ *See also* section HfC_{1-x}–SiC–Al in this table *See also* section HfC_{1-x}–Al_4C_3–SiC in Table 3.20	[526]
HfC_{1-x}–Sn	–	350	No interaction	[177, 178, 368, 627]
	–	–	Formation of Hf_2SnC ($M_{n+1}AX_n$-phase) *See also* Table 3.22 *See also* section C–Hf–Sn in Table I-2.14	
HfC_{1-x}–Ta	Vacuum	1000–1400	No interaction between powdered HfC_{1-x} and compact dense Ta	[1, 12, 13, 31, 53, 189, 195–197, 542]
	Vacuum	1400–2200	Powdered HfC_{1-x} interacts with compact dense Ta slightly	
	Vacuum	≥1400	The initiation of reaction between powdered HfC_{1-x} and compact dense Ta (exposure 5 h) *See also* section C–Hf–Ta in Table I-2.14	
HfC_{1-x}–Tc			*See* section C–Hf–Tc in Table I-2.14	
HfC_{1-x}–Th			*See* section C–Hf–Th in Table I-2.14	
HfC_{1-x}–Ti			*See* section C–Hf–Ti in Table I-2.14	
HfC_{1-x}–Tl	–	–	Formation of Hf_2TlC ($M_{n+1}AX_n$-phase) *See also* Table 3.22 *See also* section C–Hf–Tl in Table I-2.14	[177, 178]
HfC_{1-x}–U			*See* section C–Hf–U in Table I-2.14	
HfC_{1-x}–V	–	–	Stable and compatible with each other as cermet components *See also* section C–Hf–V in Table I-2.14	[552]
HfC_{1-x}–W	–	1000–1700	No interaction between powdered HfC_{1-x} and compact dense W	[1, 13, 31, 53, 186, 190, 191, 195, 198, 552]
	–	1700–2200	The interaction between powdered HfC_{1-x} and compact dense W leads to the formation of new phases	
	Vacuum	≥2000	The initiation of reaction between powdered HfC_{1-x} and compact dense W (exposure 5 h)	
	–	–	Stable and compatible with each other as cermet components	
	–	2790–2930	Eutectic HfC_{1-x}–W; the solubility of W in HfC_{1-x} is ~11 mol.% and that of HfC_{1-x} in W is ~1–3 mol.% *See also* section C–Hf–W in Table I-2.14	
HfC_{1-x}–Zn	–	550	No interaction	[368]
HfC_{1-x}–α/β-Zr	–	~2000	HfC_{1-x} is wetted readily by the molten Zr, which flows easily and reacts strongly on/with the carbide surface	[13, 192, 193, 203]
		2030	Eutectic (Hf,Zr)C_{1-x}–β-(Zr,Hf)–α-(Hf,Zr) *See also* section C–Hf–Zr in Table I-2.14	

[a]The parameters of wettability of hafnium carbide HfC_{1-x} by liquid metals are listed in Table 3.22

Table 3.20 Chemical interaction and/or compatibility of hafnium carbide HfC_{1-x} with refractory compounds at elevated, high and ultra-high temperatures (reaction systems are given mainly in alphabetical order)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	References
HfC_{1-x}–Al_4C_3	–	–	Formation of $n HfC \cdot m Al_4C_3$ compounds: $HfAl_4C_4$, $Hf_2Al_3C_4$, $Hf_2Al_4C_5$, $Hf_3Al_3C_5$, $Hf_3Al_4C_6$, $HfAl_8C_7$ and $Hf_3Al_8C_9$ *See also* section HfC_{1-x}–Al in Table 3.19 *See also* section C–Al–Hf in Table I-2.14	[513, 526, 549, 564]
HfC_{1-x}–Al_4C_3–SiC	–	–	Formation of $Hf(Al_{1-x}Si_x)_4C_4$, $Hf_2(Al_{1-x}Si_x)_4C_5$ and $Hf_3(Al_{1-x}Si_x)_4C_6$ *See also* section HfC_{1-x}–SiC–Al in Table 3.19 *See also* section HfC_{1-x}–Al_4C_3–Si in Table 3.19	[526, 549]
HfC_{1-x}–Al_4C_3–SiC–TiC_{1-x}	–	–	Formation of solid solutions $(Ti_{1-x}Hf_x)_3(Si_{1-x}Al_x)C_2$ based on Ti_3SiC_2 *See also* section HfC_{1-x}–SiC–TiC_{1-x}–Al in Table 3.19	[514]
HfC_{1-x}–Al_4C_3–$YC_{1\pm x}$/β-$YC_{2\pm x}$	–	–	Formation of $(Hf,Y)Al_4C_4$, $(Hf,Y)_2Al_4C_5$, $(Hf,Y)_3Al_4C_6$, $(Hf,Y)Al_8C_7$ and $(Hf,Y)_3Al_8C_9$ is proposed	[513]
HfC_{1-x}–$B_{4\pm x}C$	–	>1800 (?)	Formation of $HfB_{2\pm x}$ *See also* section C–B–Hf in Table I-2.14	[206–209, 571]
HfC_{1-x}–α-BN	–	–	Formation of $HfB_{2\pm x}$	[210, 211]
HfC_{1-x}–$CeN_{1\pm x}$	–	–	Terminal mutual solid solubilities between the components (?)	[31]
HfC_{1-x}–$CeP_{1\pm x}$	–	–	Terminal mutual solid solubilities between the components (?)	[31]
HfC_{1-x}–$CeS_{1\pm x}$	–	–	Terminal mutual solid solubilities between the components (?)	[31]
HfC_{1-x}–Cr_3C_{2-x}	–	–	The max. solubility of "imaginary" phase 'CrC_{1-x}' in HfC_{1-x} is ~9 mol.% and that of HfC_{1-x} in Cr_3C_{2-x} is very low (?) *See also* section C–Cr–Hf in Table I-2.14	[13, 31, 176, 236, 341]
HfC_{1-x}–Cr_3C_{2-x}–β-$Mo_{2\pm x}C$–TiC_{1-x}–δ-$TiN_{1\pm x}$–δ-$WC_{1\pm x}$	Vacuum	1350–1600	Powdered TiC_{1-x}–15% δ-$TiN_{1\pm x}$–25% δ-$WC_{1\pm x}$–8% HfC_{1-x}–8% Cr_3C_{2-x}–4% β-$Mo_{2\pm x}C$ composition was heat-treated to prepare sintered materials	[630, 631]
HfC_{1-x}–$DyN_{1\pm x}$	–	–	Monocarbonitride (cubic) continuous solid solution (complete solubility in the system?)	[31]
HfC_{1-x}–$ErN_{1\pm x}$	–	–	Monocarbonitride (cubic) continuous solid solution (complete solubility in the system?)	[31]

(continued)

Table 3.20 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	References
HfC_{1-x}–$EuN_{1\pm x}$	–	–	Monocarbonitride (cubic) continuous solid solution (complete solubility in the system?)	[31]
HfC_{1-x}–EuO	–	–	Terminal mutual solid solubilities between the components (?)	[31]
HfC_{1-x}–Fe_2O_3	–	800	The interaction initiates with the formation of α-Fe (traces)	[289]
		900	Formation of α-HfO_{2-x} and α-Fe (both in trace amounts)	
HfC_{1-x}–Fe_3C	–	–	Not (or very slightly) soluble in each other in solid state *See also* section C–Fe–Hf in Table I-2.14	[13, 341]
HfC_{1-x}–$GdN_{1\pm x}$	–	–	Monocarbonitride (cubic) continuous solid solution (complete solubility in the system?)	[31]
HfC_{1-x}–$HfB_{1\pm x}$	–	1500	Mutual solubilities of the components are ~5–7 mol.% *See also* section C–B–Hf in Table I-2.14	[31, 194, 207–209]
HfC_{1-x}–$HfB_{2\pm x}$	–	~3110–3140	Eutectic HfC_{1-x}–$HfB_{2\pm x}$; the max. solid solubility of HfC_{1-x} in $HfB_{2\pm x}$ is <2 mol.% and that of $HfB_{2\pm x}$ in HfC_{1-x} is ~10 mol.% *See also* section C–B–Hf in Table I-2.14	[12, 13, 31, 130, 194, 207–209, 571]
HfC_{1-x}–$HfN_{1\pm x}$	–	1150	Monocarbonitride (cubic) continuous solid solution (HfC_xN_z, $0.6 \le x+z \le 1$) within the homogeneity range: ~$HfC_{0.60}$–$HfN_{0.72}$–$HfN_{1.0}$–$HfC_{0.98}$	[1, 13, 31, 40, 223, 388, 407, 439, 441, 581, 629, 638]
	–	~2000	Monocarbonitride (cubic) continuous solid solution (HfC_xN_z, $0.6 \le x+z \le 1$); (complete solubility in the system, the variation of lattice parameter a, nm with compostion for $Hf(C_{1-x}N_x)$ phases ($0 \le x \le 1$) is linear: $a = 0.4640 - 0.0122x$) *See also* section C–Hf–N in Table I-2.14	
HfC_{1-x}–$HfN_{1\pm x}$ –'HfO'	–	1600	Formation of $HfC_{1-x}N_yO_z$ oxycarbonitride (cubic) phase[a] based on of HfC_{1-x}–$HfN_{1\pm x}$ continuous solid solution with homogeneity limits: $HfC_{0.75}O_{0.25}$–$HfN_{0.80}O_{0.20}$	[28, 40, 388, 406, 408, 439]
	–	2000	Formation of $HfC_{1-x}N_yO_z$ oxycarbonitride (cubic) phase based on of HfC_{1-x}–$HfN_{1\pm x}$ continuous solid solution with homogeneity limits: $HfC_{0.70}O_{0.30}$–$HfN_{0.75}O_{0.25}$ *See also* section C–Hf–N–O in Table I-2.14	

(continued)

Table 3.20 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	References
HfC_{1-x}–α/β-HfO_{2-x}	Vacuum 1.0 Pa	1000–2000	No noticeable interaction between the components (exposure 1 h)	[1, 17, 40, 213, 226, 388, 402–406, 439, 453, 471, 506, 507, 516, 523, 538, 619]
	CO, 9–133 kPa	1470–1730	Formation of $HfC_{0.93–0.97}O_{0.04–0.07}$ oxycarbide (cubic) phase	
	–	1900	Formation of $HfC_{1-x}O_y$ (HfC_{1-x}–'HfO' oxycarbide (cubic) solid solution) with the homogeneity range ~$HfC_{0.75}O_{0.25}$–$HfC_{0.55}$–$HfC_{0.99}$	
	Vacuum, CO	1900–2500	Formation of oxycarbide (cubic) phase $HfC_{0.49–0.76}O_{0.07–0.33}$ ($0.70 \leq 1-x+y \leq 1.03$) *See also* section C–Hf–O in Table I-2.14	
HfC_{1-x}–$HoN_{1\pm x}$	–	–	Monocarbonitride (cubic) continuous solid solution (complete solubility in the system, ?)	[31]
HfC_{1-x}–$LaN_{1\pm x}$	–	–	Terminal mutual solid solubilities between the components (?)	[31]
HfC_{1-x}–$LaP_{1\pm x}$	–	–	Terminal mutual solid solubilities between the components (?)	[31]
HfC_{1-x}–LaS_{1-x}	–	–	Terminal mutual solid solubilities between the components (?)	[31]
HfC_{1-x}–$LuN_{1\pm x}$	–	–	Monocarbonitride (cubic) continuous solid solution (complete solubility in the system?)	[31]
HfC_{1-x}–MgO	Vacuum 1.0 Pa	<2000	No interaction	[1, 31, 35, 212, 213, 289]
	–	2200	The initiation of reaction between powdered HfC_{1-x} and compact dense MgO (exposure 1 h); microhardness of MgO declines	
HfC_{1-x}–Mn_5C_2, Mn_7C_3	–	–	Not (or very slightly) soluble in each other in solid state *See also* section C–Hf–Mn in Table I-2.14	[341]
HfC_{1-x}–α-MoC_{1-x}	–	1400	Extended solid solution based on HfC_{1-x} with the homogeneity range from $HfC_{0.56–0.99}$ to $(Hf_{0.15}Mo_{0.85})C_{0.72–0.77}$	[13, 31, 181, 214–216, 236, 341]
	–	~1400–1700	The max. solubility of HfC_{1-x} in α-MoC_{1-x} corresponds to compos. $(Mo_{0.93–0.95}Hf_{0.05–0.07})C_{0.65–0.70}$ (the presence of Hf stabilizes α-MoC_{1-x})	
	–	~1700	Extended solid solution based on HfC_{1-x} with the homogeneity range from $HfC_{0.54–0.99}$ to $(Hf_{0.11}Mo_{0.89})C_{0.72–0.77}$	
	–	>1700–2000	Monocarbide (cubic) continuous solid solution with the homogeneity range from $HfC_{0.54–0.99}$ to $MoC_{0.64–0.71}$ (the max. C content in the mixed monocarbide phase $(Hf,Mo)C_{1-x}$ increases as Hf substitutes for Mo) *See also* section C–Hf–Mo in Table I-2.14	

(continued)

Table 3.20 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	References
HfC_{1-x}–α-MoC_{1-x}–TiC_{1-x}	–	1500–1650	Large miscibility gap connects the TiC_{1-x}–HfC_{1-x} and HfC_{1-x}–α-MoC_{1-x} boundary systems; addition of Mo to the TiC_{1-x}–HfC_{1-x} solid solution decreases the critical miscibility temp., while addition of Ti to HfC_{1-x}–α-MoC_{1-x} solid solution raises the critical temperature (no-maximum type ternary critical point) *See also* section C–Hf–Mo–Ti in Table I-2.14	[221]
HfC_{1-x}–α-MoC_{1-x}–VC_{1-x}	–	1500	Great miscibility gap between the VC_{1-x}–HfC_{1-x} and HfC_{1-x}–α-MoC_{1-x} boundary systems	[222]
	–	1500–2750	Addition of V to the HfC_{1-x}–α-MoC_{1-x} solid solution increases the critical miscibility temperature *See also* section C–Hf–Mo–V in Table I-2.14	
HfC_{1-x}–η-MoC_{1-x}	–	1700	The max. solubility of HfC_{1-x} in η-MoC_{1-x} is ~6 mol.% *See also* section C–Hf–Mo in Table I-2.14	[13, 181]
HfC_{1-x}–β-$Mo_{2\pm x}C$	–	1400–2000	The max. solubility of HfC_{1-x} in β-$Mo_{2\pm x}C$ corresponds to compos. $(Mo_{0.9}Hf_{0.1})C_{0.475-0.50}$ *See also* section C–Hf–Mo in Table I-2.14	[13, 31, 181, 236]
HfC_{1-x}–$MoSi_2$	Ar	1900–1950	Interaction in powder mixtures leads to the formation of $(Mo,Hf)_5Si_{3\pm x}$ and SiC *See also* section C–Hf–Mo–Si in Table I-2.14	[219, 220, 543, 634]
HfC_{1-x}–NbC_{1-x}	–	1200–2050	Monocarbide (cubic) continuous solid solution (complete solubility in the system) *See also* section C–Hf–Nb in Table I-2.14	[1, 13, 31, 224, 225, 341, 462, 503, 572, 632]
HfC_{1-x}–β-$Nb_{2+x}C$	–	1200–2050	The max. solubility of Hf in β-$Nb_{2+x}C$ is ~5 at.% *See also* section C–Hf–Nb in Table I-2.14	[13, 31, 225]
HfC_{1-x}–NbC_{1-x}–TaC_{1-x}			*See* section TaC_{1-x}–HfC_{1-x}–NbC_{1-x} in Table 2.22	
HfC_{1-x}–NbC_{1-x}–$ThC_{1\pm x}$	–	>2000	Terminal solid solution based on HfC_{1-x}–NbC_{1-x} monocarbide (cubic) continuous solid solution; the solubilities of $ThC_{1\pm x}$ in HfC_{1-x} and NbC_{1-x} are <5 mol.% (solubilities of HfC_{1-x} and NbC_{1-x} in $ThC_{1\pm x}$ are very low)	[31]
HfC_{1-x}–NbC_{1-x}–TiC_{1-x}	–	1900–2000	Monocarbide (cubic) continuous solid solution (complete solubility in the system)	[31, 338]
HfC_{1-x}–NbC_{1-x}–$UC_{1\pm x}$	–	1900–2050	Monocarbide (cubic) continuous solid solutions with the small miscibility gap limited by ~$(Hf_{0.4}U_{0.6})C_{1\pm x}$–~$(Hf_{0.75}U_{0.25})C_{1-x}$–~$(Hf_{0.5}Nb_{0.15-0.20}U_{0.3})C_{1-x}$ compositions *See also* section C–Hf–Nb–U in Table I-2.14	[13, 31, 338, 341]

(continued)

Table 3.20 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	References
HfC_{1-x}–NbC_{1-x}–VC_{1-x}	–	1900–2050	Monocarbide continuous solid solution with the miscibility gap limited by compos. ~$(Hf_{0.98}V_{0.02})C_{1-x}$ – ~$(Hf_{0.03}V_{0.97})C_{1-x}$– ~$(Hf_{0.2-0.3}Nb_{0.51-0.52}V_{0.2-0.3})C_{1-x}$ *See also* section C–Hf–Nb–V in Table I-2.14	[31, 224, 338]
HfC_{1-x}–NbC_{1-x}–δ-$WC_{1\pm x}$	–	1900–2000	The max. solid solubility of δ-$WC_{1\pm x}$ in HfC_{1-x}–NbC_{1-x} monocarbide (cubic) continuous solid solution corresponds to ~$(Hf_{0.6}W_{0.4})C_{1-x}$ and ~$(Nb_{0.7}W_{0.3})C_{1-x}$ compositions; practically, no solubility of HfC_{1-x} and NbC_{1-x} in δ-$WC_{1\pm x}$	[31, 338]
HfC_{1-x}–NbC_{1-x}–ZrC_{1-x}	–	1900–2000	Monocarbide (cubic) continuous solid solution (complete solubility in the system)	[31, 338, 341]
HfC_{1-x}–NbC_{1-x}–δ-NbN_{1-x}	Pure N_2	1200–2500	Monocarbonitride (cubic) continuous solid solution (complete solubility in the system)	[629]
HfC_{1-x}–δ-NbN_{1-x}	–	1600–2200	Monocarbonitride (cubic) continuous solid solution	[31, 629]
HfC_{1-x}–NbO	–	–	Terminal mutual solid solubilities between the components?	[31]
HfC_{1-x}–$NdN_{1\pm x}$	–	–	Monocarbonitride (cubic) continuous solid solution (complete solubility in the system?)	[31]
HfC_{1-x}–$Nd_2Fe_{14}B$	Pure Ar	1000	No solubility of Hf in complex boride phase, Hf carbide is stable in contact with it	[568]
HfC_{1-x}–NpC_{1-x}	–	–	Monocarbide (cubic) continuous solid solution (complete solubility in the system?)	[31]
HfC_{1-x}–$NpN_{1\pm x}$	–	–	Monocarbonitride (cubic) continuous solid solution (complete solubility in the system?)	[31]
HfC_{1-x}–$PrN_{1\pm x}$	–	–	Monocarbonitride (cubic) continuous solid solution (complete solubility in the system?)	[31]
HfC_{1-x}–PuC_{1-x}	–	–	Monocarbide (cubic) continuous solid solution (complete solubility in the system?)	[31]
HfC_{1-x}–$PuN_{1\pm x}$	–	–	Monocarbonitride (cubic) continuous solid solution (complete solubility in the system?)	[31]
HfC_{1-x}–$PuP_{1\pm x}$	–	–	Terminal mutual solid solubilities between the components (?)	[31]
HfC_{1-x}–$PuS_{1\pm x}$	–	–	Terminal mutual solid solubilities between the components (?)	[31]
HfC_{1-x}–ScC_{1-x}	–	–	Monocarbide continuous solid solution (complete solubility in the system)	[31, 340]
HfC_{1-x}–$ScN_{1\pm x}$	–	–	Monocarbonitride (cubic) continuous solid solution (complete solubility in the system?)	[31]
HfC_{1-x}–α/β-SiC	–	–	Compatible with each other within the temperature range of the thermal stability of the phases *See also* section C–Hf–Si in Table I-2.14	[1, 31, 228, 504, 633, 636]

(continued)

Table 3.20 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	References
HfC_{1-x}–α/β-Si_3N_4	–	~1400–1600	Interaction leads to the formation of Hf silicide phases and SiC	[566, 567]
HfC_{1-x}–SiO_2	–	1000–1500	Interaction leads to the formation of SiC and HfO_{2-x}	[527]
HfC_{1-x}–$SmN_{1\pm x}$	–	–	Monocarbonitride (cubic) continuous solid solution (complete solubility in the system?)	[31]
HfC_{1-x}–TaC_{1-x}			*See* section TaC_{1-x}–HfC_{1-x} in Table 2.22 *See also* section C–Hf–Ta in Table I-2.14	
HfC_{1-x}–TaC_{1-x} –$ThC_{1\pm x}$			*See* section TaC_{1-x}–HfC_{1-x}–$ThC_{1\pm x}$ in Table 2.22	
HfC_{1-x}–TaC_{1-x} –TiC_{1-x}			*See* section TaC_{1-x}–HfC_{1-x}–TiC_{1-x} in Table 2.22	
HfC_{1-x}–TaC_{1-x} –$UC_{1\pm x}$			*See* section TaC_{1-x}–HfC_{1-x}–$UC_{1\pm x}$ in Table 2.22 *See also* section C–Hf–Ta–U in Table I-2.14	
HfC_{1-x}–TaC_{1-x} –VC_{1-x}			*See* section TaC_{1-x}–HfC_{1-x}–VC_{1-x} in Table 2.22	
HfC_{1-x}–TaC_{1-x} –δ-$WC_{1\pm x}$			*See* section TaC_{1-x}–HfC_{1-x}–δ-$WC_{1\pm x}$ in Table 2.22	
HfC_{1-x}–TaC_{1-x} –ZrC_{1-x}			*See* section TaC_{1-x}–HfC_{1-x}–ZrC_{1-x} in Table 2.22	
HfC_{1-x}–α-$Ta_{2+x}C$			*See* section α-$Ta_{2+x}C$–HfC_{1-x} in Table 2.22 *See also* section C–Hf–Ta in Table I-2.14	
HfC_{1-x}–β-$Ta_{2\pm x}C$			*See* section β-$Ta_{2\pm x}C$–HfC_{1-x} in Table 2.22 *See also* section C–Hf–Ta in Table I-2.14	
HfC_{1-x}–δ-TaN_{1-x}	N_2, 3 MPa	1800	Monocarbonitride (cubic) continuous solid solution (complete solubility in the system)	[622]
HfC_{1-x}–ε-TaN_{1-x}	N_2, >0.1 MPa	1600–2000	Extended monocarbonitride (cubic) solid solutions; the max. solubility of ε-TaN_{1-x} (hexagonal) in HfC_{1-x} is ~75 mol.%	[18, 31]
HfC_{1-x}–$TbN_{1\pm x}$	–	–	Monocarbonitride (cubic) continuous solid solution (complete solubility in the system?)	[31]
HfC_{1-x}–TcC_{1-x}	–	–	Terminal mutual solid solubilities between the components (?) *See also* section C–Hf–Tc in Table I-2.14	[452, 541]
HfC_{1-x}–$ThC_{1\pm x}$	–	–	Extended monocarbide solid solutions (terminal solubility?) *See also* section C–Hf–Th in Table I-2.14	[31, 341]
HfC_{1-x}–$ThC_{1\pm x}$ –TiC_{1-x}	–	>2000	Terminal solid solution based on HfC_{1-x}–TiC_{1-x} monocarbide (cubic) continuous solid solution; the solubilities of $ThC_{1\pm x}$ in HfC_{1-x} and TiC_{1-x} are <5 and <3 mol.%, respectively (solubilities of HfC_{1-x} and TiC_{1-x} in $ThC_{1\pm x}$ are very low)	[31]

(continued)

Table 3.20 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	References
HfC_{1-x}–$ThC_{1\pm x}$–$UC_{1\pm x}$	–	>2000	The max. solubility of HfC_{1-x} in $ThC_{1\pm x}$–$UC_{1\pm x}$ monocarbide continuous solid solution corresponds to compositions ~$(Th_{0.9}Hf_{0.1})C_{1\pm x}$ and ~$(U_{0.6}Hf_{0.4})C_{1\pm x}$, respectively, and that of $ThC_{1\pm x}$ and $UC_{1\pm x}$ in HfC_{1-x} correspond to compositions ~$(Hf_{0.95}Th_{0.05})C_{1-x}$ and ~$(Hf_{0.75}U_{0.25})C_{1-x}$, respectively	[31]
HfC_{1-x}–$ThC_{1\pm x}$–VC_{1-x}	–	>2000	All the mutual solubilities between the carbides in the system are low	[31]
HfC_{1-x}–$ThC_{1\pm x}$–ZrC_{1-x}	–	2000–2100	Terminal solid solution based on HfC_{1-x}–ZrC_{1-x} monocarbide (cubic) continuous solid solution; the solubilities of $ThC_{1\pm x}$ in HfC_{1-x} and ZrC_{1-x} are <5 and <10 mol.%, respectively (solubilities of HfC_{1-x} and ZrC_{1-x} in $ThC_{1\pm x}$ are very low)	[31]
HfC_{1-x}–$ThN_{1\pm x}$	–	–	Extended monocarbonitride (cubic) solid solutions (terminal solubility?)	[31]
HfC_{1-x}–$ThP_{1\pm x}$	–	–	Terminal mutual solid solubilities between the components (?)	[31]
HfC_{1-x}–$ThS_{1\pm x}$	–	–	Terminal mutual solid solubilities between the components (?)	[31]
HfC_{1-x}–TiC_{1-x}	–	1500	The miscibility gap in the solid state – from ~$Hf_{0.1–0.2}Ti_{0.8–0.9}C_{1-x}$ to ~$Hf_{0.50–0.75}Ti_{0.25–0.50}C_{1-x}$	[13, 31, 192, 225, 229–231, 290, 457, 462, 463, 503, 575, 632]
		~1770–2050	The critical point in the miscibility gap at ~$Hf_{0.35–0.45}Ti_{0.55–0.65}C_{1-x}$	
		>1900–2000	Monocarbide continuous solid solution (complete solubility in the system); perfect homogenization after 7 h exposure *See also* section C–Hf–Ti in Table I-2.14	
HfC_{1-x}–TiC_{1-x}–$UC_{1\pm x}$	–	1900–2000	The max. solubility of $UC_{1\pm x}$ in HfC_{1-x} – TiC_{1-x} monocarbide continuous solid solution corresponds to compos. ~$(Hf_{0.75}U_{0.25})C_{1-x}$ (?) and ~$(Ti_{0.95}U_{0.05})C_{1-x}$ and that of HfC_{1-x} and TiC_{1-x} in $UC_{1\pm x}$ corresponds to compositions ~$(U_{0.6}Hf_{0.4})C_{1\pm x}$ (?) and ~$(U_{0.9}Ti_{0.1})C_{1\pm x}$, respectively (or HfC_{1-x}–$UC_{1\pm x}$ monocarbide continuous solid solution, ?) Data on the system available in the literature are contradictory	[31, 338]
HfC_{1-x}–TiC_{1-x}–VC_{1-x}	–	1900–2050	Monocarbide (cubic) continuous solid solution with the miscibility gap limited by compos. ~$(Hf_{0.98}V_{0.02})C_{1-x}$ – ~$(Hf_{0.03}V_{0.97})C_{1-x}$ – ~$(Hf_{0.14–0.17}Ti_{0.73–0.75}V_{0.09–0.12})C_{1-x}$ *See also* section C–Hf–Ti–V in Table I-2.14	[224, 338, 341]

(continued)

Table 3.20 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	References
HfC_{1-x}–TiC_{1-x}–δ-$WC_{1\pm x}$	–	1540	(Hf,W)C_{1-x}–(Ti,W)C_{1-x} monocarbide continuous solid solution with the miscibility gap limited by compos. $Hf_{0.64}Ti_{0.36}C_{1-x}$–$Hf_{0.22}Ti_{0.78}C_{1-x}$–$Hf_{0.41}Ti_{0.27}W_{0.32}C_{1\pm x}$	[31, 232, 338]
	–	1900–2000	(Hf,W)C_{1-x}–(Ti,W)C_{1-x} monocarbide (cubic) continuous solid solution (complete solubility); the max. solubility of δ-$WC_{1\pm x}$ in the continuous solid solution is corresponding to ~$(Hf_{0.6}W_{0.4})C_{1-x}$ and ~$(Ti_{0.3}W_{0.7})C_{1-x}$ compositions; practically, no solubility of HfC_{1-x} and TiC_{1-x} in δ-$WC_{1\pm x}$ *See also* section C–Hf–Ti–W in Table I-2.14	
HfC_{1-x}–TiC_{1-x}–ZrC_{1-x}	–	1900–2000	Monocarbide (cubic) continuous solid solution (complete solubility in the system)	[31, 338, 341]
HfC_{1-x}–TiC_{1-x}–δ-NbN_{1-x}	Pure N_2	1200–2500	Monocarbonitride (cubic) continuous solid solution (complete solubility in the system)	[629]
HfC_{1-x}–δ-$TiN_{1\pm x}$	–	1600–2200	Monocarbonitride (cubic) continuous solid solution	[31]
HfC_{1-x}–$TiO_{1\pm x}$	–	–	Terminal mutual solid solubilities between the components (?)	[31]
HfC_{1-x}–$UC_{1\pm x}$	–	~1730–2000	Extended monocarbide solid solutions (terminal solubility?)	[13, 31, 236, 341, 544]
		>~1730	Monocarbide (cubic) continuous solid solution (a spacing-compostion curve in the system is almost linear, complete solubility in the system?) Data on the system available in the literature are contradictory *See also* section C–Hf–U in Table I-2.14	
HfC_{1-x}–$UC_{1\pm x}$–VC_{1-x}	–	≥1900–2000	The max. solubility of HfC_{1-x} in $UC_{1\pm x}$ and VC_{1-x} corresponds to compos. ~$(U_{0.6}Hf_{0.4})C_{1\pm x}$ and ~$(V_{0.95}Hf_{0.05})C_{1-x}$, respectively; and that of $UC_{1\pm x}$ in HfC_{1-x} and VC_{1-x} corresponds to compos. ~$(Hf_{0.75}U_{0.25})C_{1-x}$ and ~$(V_{0.97}U_{0.03})C_{1-x}$, respectively; and that of VC_{1-x} in HfC_{1-x} and $UC_{1\pm x}$ corresponds to compos. ~$(Hf_{0.95}V_{0.05})C_{1-x}$ and ~$(U_{0.95}V_{0.05})C_{1\pm x}$, respectively	[31, 338]
HfC_{1-x}–$UC_{1\pm x}$–ZrC_{1-x}	–	1900–2050	Monocarbide continuous solid solution with the small miscibility gap limited by ~$(Hf_{0.4}U_{0.6})C_{1\pm x}$ – ~$(Hf_{0.75}U_{0.25})C_{1-x}$ – ~$(Hf_{0.45}Zr_{0.10}U_{0.45})C_{1\pm x}$ compositions *See also* section C–Hf–U–Zr in Table I-2.14	[13, 31, 338, 341]
HfC_{1-x}–β-UC_{2-x}	–	1800–2000	Practically, no solubility of HfC_{1-x} in α-UC_{2-x} *See also* section C–Hf–U in Table I-2.14	[13, 31]
HfC_{1-x}–UN_{1-x}	–	–	Monocarbonitride (cubic) continuous solid solution (complete solubility in the system?)	[31]

(continued)

Table 3.20 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	References
HfC_{1-x}–$UP_{1\pm x}$	–	–	Terminal mutual solid solubilities between the components (?)	[31]
HfC_{1-x}–$US_{1\pm x}$	–	–	Terminal mutual solid solubilities between the components (?)	[31]
HfC_{1-x}–UO_{2+x}	Vacuum, Ar	1930–2760	No interaction (exposure 10 min)	[1, 31, 213]
HfC_{1-x}–VC_{1-x}	–	1000	The mutual solid solubilities between HfC_{1-x} and VC_{1-x} are low	[3, 13, 31, 225, 233, 234, 236, 462, 552, 632]
	–	~2580–2650	Eutectic HfC_{1-x}–VC_{1-x}; the max. mutual solubilities of HfC_{1-x} and VC_{1-x} in each other are ~15 mol.% Data on the system available in the literature are contradictory *See also* section C–Hf–V in Table I-2.14	
HfC_{1-x}–VC_{1-x}–δ-$WC_{1\pm x}$	–	~1900–2000	The max. solubility of δ-$WC_{1\pm x}$ in cubic carbides HfC_{1-x} and VC_{1-x} corresponds to compos. ~$(Hf_{0.6}W_{0.4})C_{1\pm x}$ and ~$(V_{0.4}W_{0.6})C_{1-x}$, respectively; max. mutual solubilities of carbides HfC_{1-x} and VC_{1-x} corresponds to compos. ~$(Hf_{0.9}V_{0.1})C_{1-x}$ and ~$(V_{0.95}Hf_{0.05})C_{1-x}$ and with increasing V content in $(Hf,V)C_{1-x}$ the max. solubility of W decreases there	[31, 234, 338]
		2580	$(Hf,W)C_{1-x}$–$(V,W)C_{1-x}$ monocarbide continuous solid solution with the miscibility gap limited by compos. ~$(Hf_{0.9}V_{0.1})C_{1-x}$–~$(Hf_{0.1}V_{0.9})C_{1-x}$–~$(Hf_{0.40-0.45}V_{0.40-0.45}W_{0.15})C_{1\pm x}$; the addition of δ-$WC_{1\pm x}$ decreases the critical temp. of the miscibility gap	
HfC_{1-x}–VC_{1-x}–ZrC_{1-x}	–	1900–2050	Extended solid solution based on HfC_{1-x} – ZrC_{1-x} monocarbide (cubic) continuous solid solution; the massive miscibility gap limited by ~$(Hf_{0.98}V_{0.02})C_{1-x}$ – ~$(Hf_{0.03}V_{0.97})C_{1-x}$ – ~$(V_{0.97}Zr_{0.03})C_{1-x}$ – ~$(V_{0.08}Zr_{0.92})C_{1-x}$ compositions	[13, 31, 338, 552]
HfC_{1-x}–δ-VN_{1-x}	–	1600–2200	Monocarbonitride (cubic) continuous solid solution (?)	[18, 31]
	–	–	Terminal mutual solid solubilities between the components (?) Data on the system available in the literature are contradictory	
HfC_{1-x}–β-W_2B_{5-x}	–	2270	Eutectic HfC_{1-x}–β-W_2B_{5-x}	[521]
HfC_{1-x}–γ-WC_{1-x}	–	2550–2850	Monocarbide (cubic) continuous solid solution with the homogeneity range from $HfC_{0.54-0.99}$ to $WC_{0.60-0.66}$ (the max. C content in the mixed monocarbide phase $(Hf,W)C_{1-x}$ increases as Hf substitutes for W; the presence of Hf stabilizes γ-WC_{1-x}) *See also* section C–Hf–W in Table I-2.14	[13, 31, 191, 235, 236, 341]

(continued)

Table 3.20 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	References
HfC_{1-x}–δ-$WC_{1\pm x}$	–	1800	The max. solubility of W in HfC_{1-x} corresponds to compos. ~$(Hf_{0.7}W_{0.3})C_{1-x}$	[13, 31, 191, 235, 236, 341]
	–	2400	The max. solubility of W in HfC_{1-x} corresponds to compos. ~$(Hf_{0.4}W_{0.6})C_{1-x}$	
	–	–	Practically, no solubility of HfC_{1-x} in δ-$WC_{1\pm x}$. *See also* section C–Hf–W in Table I-2.14	
HfC_{1-x}–δ-$WC_{1\pm x}$–ZrC_{1-x}	–	1900–2000	Extended solid solution based on HfC_{1-x}–ZrC_{1-x} monocarbide (cubic) continuous solid solution; the solubilities of δ-$WC_{1\pm x}$ in HfC_{1-x} and ZrC_{1-x} are ~40 and ~30 mol.%, respectively (solubilities of ZrC_{1-x} and HfC_{1-x} in δ-$WC_{1\pm x}$ are very low)	[31, 338]
HfC_{1-x}–γ-$W_{2+x}C$	–	2660	The max. solubility of Hf in γ-$W_{2+x}C$ corresponds to compos. $(W_{0.94}Hf_{0.06})_{2.0}C_{0.99}$ *See also* section C–Hf–W in Table I-2.14	[13, 191, 341]
HfC_{1-x}–$YC_{1\pm x}$	–	–	Terminal mutual solid solubilities between the components (?)	[31]
HfC_{1-x}–$YN_{1\pm x}$	–	–	Monocarbonitride (cubic) continuous solid solution (complete solubility in the system?)	[31]
HfC_{1-x}–$YbN_{1\pm x}$	–	–	Monocarbonitride (cubic) continuous solid solution (complete solubility in the system?)	[31]
HfC_{1-x}–ZrC_{1-x}	–	from (−85)–(−100) up to 3400	Monocarbide (cubic) continuous solid solution, complete solubility in the system (data on the higher melting points of $(Hf,Zr)C_{1-x}$ in comparison with individual HfC_{1-x} and ZrC_{1-x} are not confirmed); the critical point of the miscibility gap is corresponding to (−85)–(−100) °C and ~$Hf_{0.50}Zr_{0.50}C_{1-x}$ composition (theoretically calculated)	[13, 31, 46, 192, 193, 225, 229, 290, 341, 454, 456, 457, 465, 565, 574–576, 632]
	–	2400–2800	Sintering process of the equimolar mixture of components was studied *See also* section C–Hf–Zr in Table I-2.14	
HfC_{1-x}–$ZrN_{1\pm x}$	–	–	Monocarbonitride (cubic) continuous solid solution (?)	[31]
HfC_{1-x}–$ZrP_{1\pm x}$	–	–	Terminal mutual solid solubilities between the components (?)	[31]
HfC_{1-x}–β/γ-ZrO_{2-x}	Vacuum	2000–2300	No interaction between dense compact HfC_{1-x} and powdered β/γ-ZrO_{2-x}	[1, 35, 213, 289]
	–	>2200	The initiation of reaction between powdered HfC_{1-x} and compact dense γ-ZrO_{2-x} (exposure 1 h)	

[a]Practically, due to the serious difficulties in manufacturing transition metal carbide materials non-contaminated in any degree by O and N, all the materials labeled in literature as HfC_{1-x} with the appropriate certification more likely would have to be considered as $HfC_{1-x}N_yO_z$ with low or very low values of y and z indexes

Table 3.21 Chemical interaction of hafnium monocarbide HfC_{1-x} with gaseous media at elevated, high and ultra-high temperatures (reaction systems are given in alphabetical order)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	References
HfC_{1-x}–CO	CO	1400	The equilibrium pressure of CO (p_{CO}) of the reaction HfC + 2CO ↔ α-HfO_2 + 3C is ~0.01 MPa	[28, 40, 242, 243, 388, 453, 523, 538]
	CO	1500–1640	p_{CO} (1500 °C) = ~0.02 MPa, p_{CO} (1600 °C) = ~0.05 MPa, p_{CO} (1640 °C) = ~0.1 MPa	
	CO	~1630	At pressures <1.33×10^{-4} Pa the surface properties of single crystal HfC_{1-x} are stable	
	CO	~1600–1900	Formation of oxycarbide phases $HfC_{1-x}O_y$ (extended substitution solid solution based on HfC_{1-x} with approximate homogeneity limits: ~$HfC_{0.75}O_{0.25}$–$HfC_{0.55}$–$HfC_{0.99}$) *See also* section C–Hf–O in Table I-2.14	
HfC_{1-x}–CO_2	CO_2	–	Formation of oxycarbide phases $HfC_{1-x}O_y$ (extended substitution solid solution based on HfC_{1-x} with approximate homogeneity limits: ~$HfC_{0.75}O_{0.25}$–$HfC_{0.55}$–$HfC_{0.99}$) and subsequent formation of oxide scales	[28, 40, 237, 243, 388, 453, 523, 538]
	CO_2, <10^{-4} Pa	~1630	In the range of low pressures CO_2 the surface properties of single crystal HfC_{1-x} are stable *See also* section C–Hf–O in Table I-2.14	
HfC_{1-x}–F_2	F_2, 0.1 MPa	25	Arc-melted $HfC_{0.95}$ is resistant to F_2	[245]
	F_2–He	~300	Detectable highly exothermic reaction starts (at fluorine partial pressure $p_{F2} \geq 0.37$ kPa); the surface temperature of the samples increase rapidly	
	F_2–He	~300–400	Formation of solid HfF_4, which is not adherent to the carbide surface; so, the $HfC_{0.95}$ surface is practically clean (p_{F2} = 0.37 kPa)	
	F_2–He	~400–500	Formed solid HfF_4 is more adhered to the arc-melted $HfC_{0.95}$ surface, but it is abrupt crumbled (p_{F2} = 0.37 kPa)	
	F_2 (0.4–7.0 kPa), He (~100 kPa)	~550–900	Due to the volatilization of outer fluoride layers, arc-melted $HfC_{0.95}$ loses its mass (the reaction rate k is linear); k, mg cm^{-2} s^{-1}, varies from 0.03 (565 °C, p_{F2} = 0.37 kPa; incomplete volatilization of HfF_4) to 0.8 (820 °C, p_{F2} = 7.0 kPa); at temperatures >600–650 °C apparent activation energy $E \approx 0$ and reaction rate is directly proportional to p_{F2}	

(continued)

Table 3.21 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	References
HfC_{1-x}–H_2	H_2	–	Formation of carbohydride phases HfC_xH_y (x = 0.50–0.58, y = 0.42–0.43, including Hf_2CH_x) via hydrogen dissolution in non-stoichiometric phases (dehydrogenation leads to the starting compositions again)	[31, 154, 244, 467–469, 520]
	H_2	2230–2730	HfC_{1-x} is nonreactive with H_2; it also has low permeability to H_2	
	H_2	<2480	HfC_{1-x} is resistant to H_2	
	H_2	>2760	HfC_{1-x} interacts with H_2 *See also* section C–H–Hf in Table I-2.14	
HfC_{1-x}–H_2O	H_2O	400–500	No interaction	[154, 470, 489]
	H_2O, He	700–800	Hydrolysis of HfC_{1-x} materials in superheated steam leads to the evolution of H_2, CO, CO_2 (delayed in release) and small amounts of CH_4, and precipitation of elemental C and C_nH_m wax (primary products), which are converted thereafter into CO_2 and H_2; a large amount of H_2 evolves at the earlier stage while CO_2 continues to evolve with some amount of H_2 even after the ceasing of CO evolution	
HfC_{1-x}–N_2	N_2	20–1000	For wafers heated at 10 K min^{-1} rate the mass gain started in a range from 180 to 590 °C and achieved up to 1–6% in total	[1, 13, 29, 31, 35, 92, 153, 154, 223, 246, 365, 379, 381, 441, 527]
	N_2, ≤30 MPa	1100–1400	No interaction	
	N_2	1150	Formation of monocarbonitride (cubic) continuous solid solution with the homogeneity range within ~$HfC_{0.60}$–$HfN_{0.72}$–$HfN_{1.0}$–$HfC_{0.98}$ compositions	
	N_2 flow	1500	No interaction, no mass change	
	N_2, 0.1–30 MPa	1500–1800	The formation of various monocarbonitride HfC_yN_z ($0.6 \leq x + z \leq 1.0$) phases Data available in literature are controversial *See also* section C–N–Hf in Table I-2.14	
HfC_{1-x}–O_2[a,b]	O_2	20–25	At high O coverages single crystal $HfC_{0.98}$ (100) surface reacts with the depletion of C atoms and formation of disordered oxide layers (finally – HfO_{2-x}-like layers)	[1, 35, 217, 237, 238, 248–262, 388, 394–396, 400, 439, 453, 471–476, 522, 523, 535, 538, 563, 619]
	O_2–Ar	400–420	Powdered (grain size – ~1 μm, BET surface area – 1.0 m^2 g^{-1}) near-stoichiometric HfC_{1-x} starts to be oxidized (at partial pressures p_{O2} = 0.5–40 kPa) with formation of $HfC_{1-x}O_y$, α-HfO_{2-x} (monoclinic) and C (amorphous)	
	Air	410–435	The initiation of oxidation reaction for powdered $HfC_{0.76}O_{0.07}$, $HfC_{0.78}O_{0.26}$ and $HfC_{0.60}$	

(continued)

Table 3.21 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	References
	Air	435–900	In this temperature range the oxidation of powdered $HfC_{0.60}$ with heating rate 10 K min^{-1} leads to 13% total mass gain and maximal mass gain rate of 0.14 mg s^{-1}; a certain amount of combined carbon was revealed in the final oxidation products due to the incomplete decomposition of ~HfO_2C intermediate phase[c]	
	O_2–Ar	480–600	Powdered (grain size – ~1 μm, BET surface area – 0.33 m^2 g^{-1}) $HfC_{0.98}$ oxidizes at p_{O2} = 4–16 kPa through the initial stage of oxycarbide phase $HfC_{1-x}O_y$ formation and subsequent stages of O-diffusion-controlled in α-HfO_{2-x} (early) and phase-boundary-control-led (later) processes (both described by apparent activation energy E = 195 ± 15 kJ mol^{-1})	
	O_2–Ar	600–900	The oxidation of single crystal $HfC_{0.96}$ with faces parallel to the (100) plane at p_{O2} = 2–8 kPa is accompanied with the formation of bilayered C containing α-HfO_{2-x} and described by linear kinetics with apparent activation energy (>700 °C) E = 160 kJ mol^{-1}	
	Air	600–1200	The oxidation of compact (porosity 1–2%) near-stoichiometric HfC_{1-x} results in the formation of oxycarbide phase $HfC_{1-x}O_y$, C (amorphous) and oxide (α- and β-HfO_{2-x}) phases; occurrence of $HfN_{1\pm x}$ and metal Hf (?) was detected too	
	O_2–Ar	700–1500	Single crystal $HfC_{0.96}$ with faces parallel to the (200) plane at p_{O2} = 0.08–80.0 kPa oxidizes through the formation of oxycarbide $HfC_{1-x}O_y$, C and α-HfO_{2-x} with preferred (200), (020) or (002) orientation; the C containning oxide scale consists of compact/poreless (inner, ~23–25 at.% C) and porous (outer, ~6–11 at.% C) layers	
	O_2, 0.1 MPa	1000	Disintegration of dense zone-refined $HfC_{0.97}$ (contents: O – 0.003%, N – 0.031%, Fe – 0.005%, Zr – 0.035%) samples into several pieces within 3 min	
	Air	1100–1400	Severe oxidation and disintegration of HfC_{1-x} materials	
	Air	1200	The oxidation mass gains of hot-pressed (1–2% porosity) HfC_{1-x} are 0.65 and 1.15 kg m^{-2} for 1 h and 2 h exposure, respectively	

(continued)

Table 3.21 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	References
	O_2, 2–100 kPa	1200–1800	The oxidation rate constants decrease with temperature growth; CO is formed at the carbide-oxide interface and oxidized by incoming O_2 to form CO_2 in the outer layer of oxide scale where the morphology abruptly changed in comparison with the inner layer of scale	
	Air	1200–1800	The oxidation kinetics of hot-pressed (~10% porosity) near-stoichiometric HfC_{1-x} is controlled by gaseous diffusion via pores in the oxide (parabolic growth of the scale)	
	O_2–Ar	1400–2100	CVD-prepared $HfC_{\sim 0.5}$ film oxidizes by 93% Ar – 7% O_2 gas flow (20 cm s^{-1}) through the formation of 3 layers: residual carbide with dissolved O ($HfC_{1-x}O_y$), dense-appearing oxide containing carbon ($HfO_{2-x}C_y$) and porous outer oxide layer (α-HfO_{2-x})	
	Air	1500	The oxidation mass gains of the bulk samples of $HfC_{0.67}$, $HfC_{0.82}$ and $HfC_{0.98}$ are 1, 14 and 21 mg cm^{-2}, respectively (0.25 h exposure); morphological observations show the formation of a "maltese cross" protective oxide scale structure	
	O_2 (0.6–1.5 kPa), He (100 kPa, flow rate ~2 cm^3 s^{-1})	1500–2000	The oxidation of dense zone-refined $HfC_{0.97}$ (contents: O – 0.003%, N – 0.031%, Fe – 0.005%) is linear in character and preferential along grain boundaries (the average oxidation mass gain is ~15–17 mg cm^{-2} for 1–1.5 h exposure at T = 1500–1600 °C)	
	Air	1800–2200	The oxidation kinetics of hot-pressed (~10% porosity) near-stoichiometric HfC_{1-x} is controlled by bulk (ambipolar) diffusion of O atoms (parabolic growth of the scale); the scale substantially densifies, yet remaining sufficiently porous to provide CO release	
	O_2/C_2H_2 torch	~3000	The oxidation of CVD-prepared HfC_{1-x} coating by gas flow (pressure O_2/C_2H_2 – 0.4/0.1 MPa, flow rate – 0.7–2 dm^3 s^{-1}) leads to the formation of $HfO_{2-x}C_y$ interlayer and HfO_{2-x} scale (β-HfO_{2-x} formed on the surface transforms to α-HfO_{2-x} upon cooling to room temperature); the oxide scale is composed of molten and porous (with micro-cracks and microholes as a result of CO/CO_2 release) layers	
			See also section C–O–Hf in Table I-2.14	

[a]Fine hafnum carbide powders are pyrophoric [247]

[b]For near-stoichiometric HfC_{1-x} the value of the Pilling-Bedworth ratio (based on the density of the low-temperature (monoclinic) α-HfO_{2-x} oxide phase) $\alpha = M_O d_C / M_C d_O = 1.44$, where M_O is molecular mass of the oxide phase formed on the oxidation of 1 mol of carbide phase, M_C is molecular mass of carbide phase, d_C and d_O are the densities of carbide and oxide phases, respectively [394]

[c]By analogy with the phase ZrO_2C reported by Guerlet and Lehr [401]

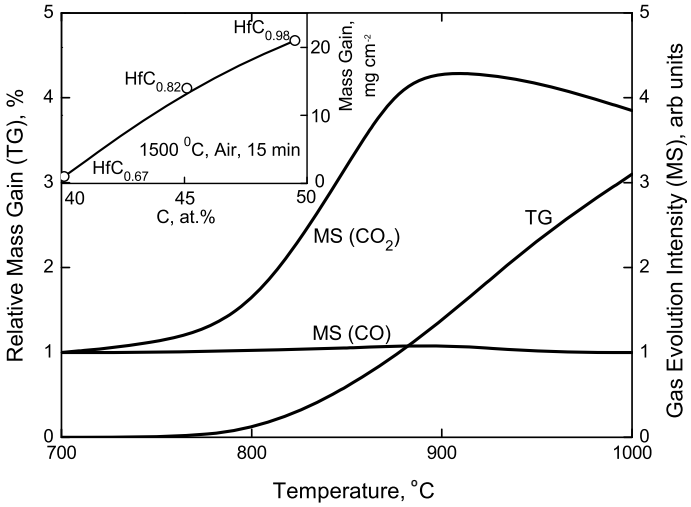

Fig. 3.22 Non-isothermal oxidation kinetics curves (heating rate – 0.17 °C s^{-1}) for ~60 mg samples of single crystal $HfC_{0.96}$ with faces parallel to the (100) plane exposed to a flowing atmosphere (p = 20 kPa) of the mixed gases of 20 vol.% O_2 and 80 vol.% Ar: TG – thermogravimetric (mass increase begins at ~770 °C) and MS – mass-spectrometric (CO_2 gas evolution begins at ~770 °C with a maximum at ~890 °C, CO gas was hardly detected); the complete oxidation and conversion of carbide to oxide in accordance with the reaction $HfC_{0.96} + 1.96O_2 \rightarrow HfO_2 + 0.96CO_2$ is corresponding to the relative mass gain of 10.5% [237] (*Inset* – the variation of oxidation mass gain with carbon content for HfC_{1-x} materials with different deviations from the stoichiometry [394–396])

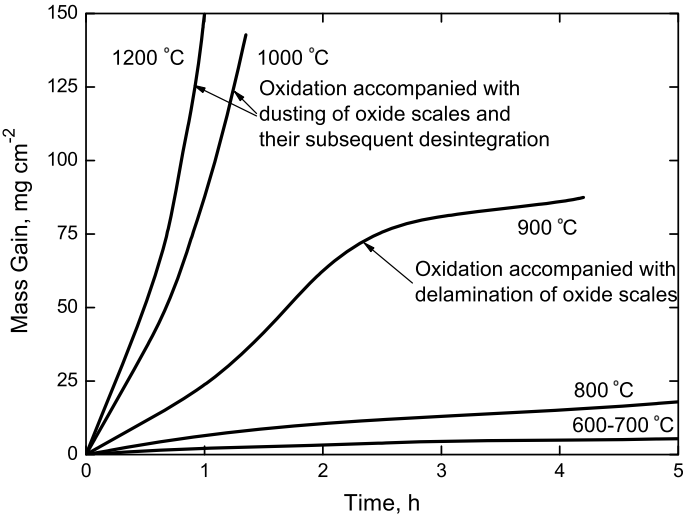

Fig. 3.23 Isothermal oxidation kinetics curves for 8 mm diameter disk-like samples of hot-pressed and subsequently annealed near-stoichiometric hafnium monocarbide HfC_{1-x} materials (porosity 1–2%) in air [238–241]

Table 3.22 The parameters of wettability of hafnium monocarbide HfC_{1-x} phases with some liquid metals and alloys (melts)[a]

Melt (purity)	Atmosphere	Temp., °C	Time, s	γ_{l-g}, mJ m^{-2}	W_a, mJ m^{-2}	W_m^b, kJ mol^{-1}	θ, degree	References
				$HfC_{0.99}$				
Co[c]	–	1500	900	1805	3260	170	36 ± 2	[1, 263]
Fe[d]	Vacuum	1550	900	1780	2870	150	52 ± 2	[1, 263]
Ni[e]	–	1450	900	1700	3400	177	28 ± 2	[1, 263]
				$HfC_{\sim 1.0}$				
Al (99.97%)	Vacuum	900	900	914	140	7.1	148 ± 3	[1, 35, 265, 272]
Al (99.97%)	Vacuum	1000	900	914	155	–	146 ± 3	[1, 35, 265, 272]
Bi (99.999%)	Vacuum	320	900	390	60	3.1	148 ± 2	[1, 35, 271, 272]
Co	Vacuum	1420	300	1910	3380	–	40	[1, 266, 271, 272]
Co[c]	–	1500	1200	1805	3265	170	36	[1, 263, 267, 268]
Co (99.98%)	Ar	1500	1200	1805	3190	165	40	[1, 265, 345]
Co	Ar	1550	–	–	–	–	40	[35, 271, 272, 345]
Co	Vacuum	1550	–	–	–	–	36	[35, 271, 272]
Cu	Vacuum	1100	–	–	–	–	134	[35, 266, 271, 626]
Cu (99.99%)	Vacuum	1130	900	1351	316	16	140 ± 0.25	[1, 265, 272, 345]
Cu[f]	Vacuum, 1.3 mPa	1150	3600	610	–	–	134	[625]
Cu (Hf – 2.75%)[f]	Vacuum, 1.3 mPa	1150	3600	–	–	–	146	[625]
Cu (Hf – 2.75, V – 0.8%)[f]	Vacuum, 1.3 mPa	1150	3600	–	–	–	105	[625]
Cu	Vacuum	1200	–	–	–	–	131	[35, 266, 271, 272]
Fe	Vacuum	1490	60	1900	3245	–	45	[1, 266, 269, 272]
Fe (C-4.0, Mn-1.4, Si-2.5, S-0.1%)[g]	Pure Ar	1400–1450	900	–	–	–	132	[1, 270, 272]
Fe (C-3.1, Mn-0.4, Si-1.6, S-0.02%)[g]	Pure Ar	1400–1450	900	–	–	–	122	[1, 270, 272]

(continued)

Table 3.22 (continued)

Melt (purity)	Atmosphere	Temp., °C	Time, s	γ_{l-g}, mJ m^{-2}	W_a, mJ m^{-2}	W_m^b, kJ mol^{-1}	θ, degree	References
Fe (C-2.7, Mn-0.4, Si-1.2, S-0.2%)[g]	Pure Ar	1400–1450	900	–	–	–	145	[1, 270, 272]
Fe (C-2.6, Mn-1.0, Si-1.0, Cr-0.3%)[g]	Pure Ar	1400–1450	900	–	–	–	140	[1, 270, 272]
Fe (C-1.0, Cr-1.4, Mn-0.4, Si-0.3%)[g]	Pure Ar	1500–1550	900	–	–	–	132	[1, 270, 272]
Fe (C-1.0, Cr-1.4, Mn-1.1, Si-0.6%)[g]	Pure Ar	1500–1550	900	–	–	–	145	[1, 270, 272]
Fe (C-0.8, Mn-0.2, Si-0.2, S-0.03%)[g]	Pure Ar	1500–1550	900	–	–	–	148	[1, 270, 272]
Fe (C-0.2, Cr-14.5, Ni-2.7, Mn-0.5%)[g]	Pure Ar	1500–1550	900	–	–	–	128	[1, 270, 272]
Fe[d]	Vacuum	1550	900	1780	2870	149	52	[1, 263, 271, 272]
Fe (99.999%)	Ar	1550	900	1780	270	14	148	[1, 265, 271, 272]
Ga (99.99999%)	Vacuum, Ar	800	900	707	115	6.1	147 ± 0.3	[1, 35, 271, 272]
Ge (99.9999%)	Vacuum	1000	900	600	140	7.3	140 ± 0.3	[1, 35, 271, 272]
In (99.9995%)	Vacuum	250	900	559	90	4.6	147 ± 0.6	[1, 35, 271, 272]
Mn (99.8%)	Vacuum	1300	900	1750	2115	110	78	[1, 265, 345]
Mn (99.8%)	Ar	1300	900	1750	1870	97	86	[1, 265, 345]
Mo	He	2610	–	–	–	–	14	[271]
Ni	Vacuum	1380	300	1810	3475	–	23	[1, 35, 266, 272]
Ni[e]	Vacuum, Ar	1450	900	1700	3200	170	28	[1, 263, 271, 272]
Ni (99.99%)	Ar	1450	900	1700	3060	177	37	[1, 265, 345]
Pb (99.98%)	Vacuum	400	900	480	64	3.3	150 ± 1	[1, 35, 271, 272]
Sb (99.999%)	Vacuum	700	900	384	137	7.1	130 ± 1	[1, 35, 271, 272]

(continued)

Table 3.22 (continued)

Melt (purity)	Atmosphere	Temp., °C	Time, s	γ_{l-g}, mJ m^{-2}	W_a, mJ m^{-2}	W_m^b, kJ mol^{-1}	θ, degree	References
Si	Vacuum	1500	900	860	1650	84	23 ± 2	[1, 35, 272]
Sn (99.999%)	Vacuum	300	900	554	61	3.1	153 ± 2	[1, 35, 271, 272]
Tl (99.999%)	Vacuum	400	900	490	163	8.4	132 ± 3	[1, 35, 271, 272]

[a]The parameters of wettability are given in accordance with Young-Dupré equation $W_a = \gamma_{l\text{-}g} \times (1 + \cos\theta)$ and Young's equation $\gamma_{s\text{-}l} = \gamma_{s\text{-}g} - \gamma_{l\text{-}g}\cos\theta$, where W_a is the work of adhesion, $\gamma_{l\text{-}g}$ is the liquid-vapour interfacial energy (surface tension), $\gamma_{s\text{-}l}$ is solid-liquid interfacial energy, $\gamma_{s\text{-}g}$ is the solid-vapour interfacial energy and θ is the wetting contact angle [1]; compositions of melts are given in mass (weight) percentage

[b]$W_m = W_a(M/d)^{2/3}N_A^{1/3}$, where W_m is the molar work of adhesion, M is the molecular mass and d is the density of chemical compound, N_A is the Avogadro constant [264]

[c]$\gamma_{s\text{-}l}$ = 1500 mJ m^{-2}

[d]$\gamma_{s\text{-}l}$ = 1860 mJ m^{-2}

[e]$\gamma_{s\text{-}l}$ = 1450 mJ m^{-2}

[f]Sintered materials (porosity – ≤2%, contents: non-combined C – 0.20%, O – 0.06%, N – 0.10%, Cr – 0.30%, Zr – 0.30%)

[g]Sintered materials (porosity – 3–6%)

compounds – in Table 3.20 and with gaseous media – in Table 3.21. The data on the oxidation resistance of HfC_{1-x} listed there are also accompanied by the graphic information in Figs. 3.22 and 3.23. Data on the catalytic activities of HfC_{1-x} are given in several works [1, 282, 286, 347–356, 466, 495, 542]; the examples of its electrochemical behaviour are described in [291, 292, 357, 374]. The parameters of wettability of HfC_{1-x} with some liquid metals (melts) are listed in Table 3.22, the diffusion rates for the systems containing HfC_{1-x} are presented in species pairs within the various ranges of temperatures in the Table 3.23. The characters of chemical interaction of hafnium monocarbide HfC_{1-x} with some common chemicals (acids, alkalies and salts in aqueous solutions) are summarized in Table 3.24.

Table 3.23 Diffusion rates and related parameters in the systems containing hafnium, carbon and hafnium monocarbide phases at various temperatures[a, b]

Species pair	Temperature dependence of the diffusion coefficient (diffusivity) $D = D_0\exp[(-E_A/R)/T]$, $cm^2\ s^{-1}$	Temperature range, K (°C)	Remarks on materials characteristics and measurement method	References
C → α-Hf	$74\exp(-37{,}600/T)$	1390–2030 (1120–1760)	Polycrystalline Hf (1.5 wt.% Zr), residual activity method using ^{14}C radioactive diffusant	[1, 14, 35, 273–275]
	$6.1 \times 10^{-5}\exp(-12{,}000/T)$, ?	1370–1570 (1100–1300)	Calculation based on the growth of HfC_{1-x} phase layer	[1]
C → β-Hf	$4.2 \times 10^{-2}\exp(-20{,}100/T)$	2030–2400 (1760–2130)	Polycrystalline Hf (1.5 wt.% Zr), residual activity method using ^{14}C radioactive diffusant	[1, 14, 15, 35, 273]
	$0.8\exp(-25{,}400/T)$	2070–2370 (1800–2100)	Polycrystalline Hf (3 wt.% Zr), diffusion couple method with determination of $c \sim x$ curves by calculation from an analytical solution	[274, 275]
C → HfC_{1-x}	$63\exp[(-65{,}600 \pm 6600)/T]$	2470–3070 (2200–2800)	Polycrystalline (hot-pressed) HfC_{1-x} ($x = 0.03$, $Hf(C_{0.95}N_{0.04}O_{0.01})_{0.97}$, 100–300 μm grain size, porosity <5%, $O + N \leq 0.05\%$), ^{14}C radiotracer	[4, 35, 277, 278, 358, 360, 361, 547]
	$4.7 \times 10^{4}\exp[(-64{,}500 \pm 3300)/T]$	1860–2270 (1590–2000)	Polycrystalline (hot-pressed) HfC_{1-x} ($x \approx 0$, porosity – 5%, non-combined C – 0.9%), ^{14}C radiotracer	[4, 14, 15, 278, 280, 358]
	$2.62\exp(-48{,}000/T)$	2570–3270 (2300–3000)	Parameters of temperature variation of HfC_{1-x} layer growth rate constant upon C saturation of liquid Hf in graphite crucible	[14, 15, 279]
	$8.6\exp(-51{,}200/T)$	2570–3270 (2300–3000)	Parameters of reaction-chemical diffusion, C saturation of liquid Hf in graphite crucible; metallography method	[14, 15, 279, 281, 369]
Hf → HfC_{1-x}	$1.48 \times 10^{8}\exp(-92{,}000/T)$	–	Values calculated on the diffusion theory basis	[1, 35, 276]
	$\sim\exp(-75{,}200/T)$	–	Estimated on the basis of LeClaire's relation (connected with phase melting point)	[369]

(continued)

Table 3.23 (continued)

[a]The chemical diffusion coefficient concentration and temperature dependencies can be represented by an equation of the form $D' = D'_0\exp(\beta x)\exp[(-E'_A/R)/T]$, where β is the concentration factor and x is the value of index in HfC_{1-x} formula (for the ranges of x = 0.01–0.50 and T = 2330–3030 °C the recommended values are D'_0 = 3.94 $cm^2\ s^{-1}$, β = 9.2 and E'_A = 477 kJ mol^{-1} [281]), and the discrepancy between chemical and self-diffusivities may be estimated on the basis of the thermodynamical relationship: $D' = D \times [1 + (d\ln\gamma_C/d\ln N_C)]$, where γ_C is the activity coefficient of carbon and N_C is the atomic fraction of carbon (evaluated at 2000 °C in HfC_{1-x}: at x = 0.01, $d\ln\gamma_C/d\ln N_C$ = 141 and at x = 0.20, $d\ln\gamma_C/d\ln N_C$ = 15 [359]); the approximate values of apparent activation energy for some diffusion controlled processes in HfC_{1-x}: **(a)** grain growth (particle coarsening) – 210 ± 5 kJ mol^{-1} ($HfC_{\sim 1.0}$, mean particle size 45–50 nm, in arc-melted W – 3.6 mas.% Re alloy, 1900–2300 °C) [481]; **(b)** powder hot-pressing densification – 160 kJ mol^{-1} ($HfC_{0.71}$, mean particle size 10–20 μm, 2400–2600 °C) [477], 170 kJ mol^{-1} ($HfC_{0.80}$, mean particle size 10–20 μm, 2400–2600 °C) [477], 190 kJ mol^{-1} ($HfC_{0.91}$, mean particle size 10–20 μm, 2400–2600 °C) [477], 215 kJ mol^{-1} ($HfC_{\sim 1.0}$, mean particle size 10–20 μm, 2400–2600 °C) [477]; data on creep – *see* Sect. 3.4 (Table 3.13), *see also* Sect. 3.5 (Table 3.18)

[b]By the studies of diffusion in HfC_{1-x}–TaC_{1-x} powder blends annealing at 1900 °C (exposure time – from 5 to 200 h) it was found that Hf atoms have a higher intrinsic diffusivity compared with Ta atoms [397]

Table 3.24 The interaction of near-stoichiometric hafnium monocarbide HfC_{1-x} materials with some common chemical reagents in aqueous solutions [1, 32, 35, 36, 45, 64, 247, 283–285, 291, 292, 366, 370, 482]

Reagent, formula (density or concentration of aqueous solution)[a]	Treatment conditions		Character of interaction[b]
	Temperature, °C	Exposure time, h	
HCl (1:1)	20	24	Decomposes up to ~4%
	110–112	2	Decomposes up to ~57%
HCl (d = 1.19)	20	24	No decomposition
	100–120	2	No decomposition
H_2SO_4 (1:4)	20	24	No decomposition
	110–116	2	Decomposes up to ~10–12%
H_2SO_4 (d = 1.84)	20	24	No decomposition
	264–268	1	Decomposes completely (16.1% amorphous C); composition of released gases: CH_4 64.2, CO_2 31.8 and H_2 4.0 mol.%[c]
	290–300	1	Decomposes completely (6.3% amorphous C); composition of released gases: CH_4 16.2, CO_2 81.3 and H_2 2.4 mol.% (solution is transparent finally)[c]
HNO_3 (1:1)	20	24	Decomposes up to ~25%
	100–105	2	Decomposes completely (no precipitated C, solution is transparent)[d]
HNO_3 (d = 1.43)	20	24	Decomposes up to ~40%
	110–120	0.5	Decomposes completely with the formation of soluble nitrates (no precipitated C, solution is transparent); composition of released gases: CH_4 (traces), CO_2 95–97 and H_2 2–3 mol.% (N oxides are not taking into account)[d]
H_3PO_4 (1:3)	20	24	Decomposes up to ~2–3%
	≥100	2	Decomposes up to ~10%
H_3PO_4 (d = 1.70)	20	24	Decomposes up to ~3%
	110	2	Decomposes up to ~10%
	230–240	1.5	Decomposes up to ~77%; composition of released gases: CH_4 62.1, CO_2 35.2 and H_2 3.0 mol.%[e]
	250	1	Decomposes completely with the formation of soluble phosphate $HfO(H_2PO_4)_2$; composition of released gases: CH_4 77.2, CO_2 19.3 and H_2 3.4 mol.% (dissolution proceeds very energetically and mainly finishes in 30–40 min)[e]

(continued)

Table 3.24 (continued)

Reagent, formula (density or concentration of aqueous solution)[a]	Treatment conditions		Character of interaction[b]
	Temperature, °C	Exposure time, h	
	250–270	1.5	Decomposes completely with the formation of soluble phosphate $HfO(H_2PO_4)_2$; composition of released gases: CH_4 69.3, CO_2 28.9 and H_2 1.7 mol.% (resultant solution is transparent, on heavily diluting with H_2O the amorphous precipitates appear, which after heating transform into pyrophosphate $Hf_2P_2O_7$)[e]
$HClO_4$ (d = 1.67)	20	24	Decomposes up to ~3%
	≥100	2	Decomposes up to ~98%
$H_2C_2O_4^f$ (saturated solution)	20	24	Decomposes up to ~2–3%
	≥100	2	Decomposes up to ~10%
Organic acids	110	1	No decomposition
NaOH (10%)	20	24	No decomposition
	108	2	Decomposes up to ~2%
NaOH (20%)	20	24	No decomposition
	108–110	2	Decomposes up to ~1–2%
H_2O_2 (30%)	110	1	Practically, no decomposition
NaCl (3%)	–	–	The oxide film forms on the HfC_{1-x} anode under electrolysis at potentials between −0.2 and +1.9 V; the film consists of the upper layer formed in polymolecular chemosorption of O_2 and Cl_2 gases followed by the HfO_{2-x} + C layer and inner oxycarbide HfC_xO_y layers
Na_2SO_4	–	–	Under electrolysis HfC_{1-x} has a wider passivity region than metal Hf and dissolves anodically faster
$NaNO_3$	–	–	Under electrolysis HfC_{1-x} has a wider passivity region than metal Hf and dissolves anodically faster
NH_4F (5%)	–	–	Decomposes up to ~23%
$(NH_4)_2S_2O_8$ (25%)	–	–	Decomposes completely
3HCl (d = 1.19) + HNO_3 (d = 1.43)	20	24	Decomposes up to ~86%
	106–110	2	Decomposes completely
3HCl (1:1) + HNO_3 (1:2)	110	1	Decomposes almost completely
HNO_3 (d = 1.43) + H_2SO_4 (d = 1.84)	20	24	Decomposes completely
	110	2	Decomposes completely
H_2SO_4 (1:4) + H_3PO_4 (1:3)	20	24	Decomposes almost completely
	≥ 150	2	Decomposes completely
H_2SO_4 (d = 1.84) + H_3PO_4 (d = 1.70)	20	24	Decomposes up to ~1–3%
	250	2	Decomposes almost completely with the deposition of salts
H_2SO_4 (1:4) + $H_2C_2O_4^f$ (saturated solution)	20	24	No decomposition
	≥100	2	Decomposes up to ~5%

(continued)

Table 3.24 (continued)

Reagent, formula (density or concentration of aqueous solution)[a]	Treatment conditions		Character of interaction[b]
	Temperature, °C	Exposure time, h	
H_2SO_4 (d = 1.84) + $H_2C_2O_4$[f] (saturated solution)	20	24	Decomposes up to ~2%
	≥100	2	Decomposes up to ~34–37%
HCl (d = 1.19) + H_2O_2 (30%)	110	1	Decomposes almost completely
HCl (d = 1.19) + $(NH_4)_2S_2O_8$ (25%)	–	–	Decomposes completely
HCl (d = 1.19) + Br_2 (HBrO, HBr) (saturated solution)	–	–	Decomposes almost completely
H_2SO_4 (1:4) + H_2O_2 (30%)	110	1	Decomposes completely
H_2SO_4 (1:4) + $(NH_4)_2S_2O_8$ (25%)[g]	110	1	Decomposes completely
H_2SO_4 (1:1) + $(NH_4)_2S_2O_8$ (25%)	110	1	Decomposes completely
$4HNO_3$ (d = 1.43) + HF (d = 1.13)	20	24	Decomposes completely
	110	1	Decomposes completely
H_2O_2 (30%) + $H_2C_2O_4$[f] (50%)	110	1	Decomposes completely
H_2O_2 (30%) + $H_2C_4H_4O_6$[h] (50%)	20	24	No decomposition
	110	1	Decomposes up to ~3%
H_2O_2 (30%) + $H_3C_6H_5O_7$[i]	110	1	Decomposes up to ~2%
H_2O_2 (30%) + $H_4C_{10}H_{12}N_2O_8$[j] (saturated solution)	110	1	Decomposes up to ~3%
H_2O_2 (30%) + NH_4F	110	1	Decomposes completely
$(NH_4)_2S_2O_8$ + $H_4C_{10}H_{12}N_2O_8$[j]	–	–	Decomposes up to ~28%
$(NH_4)_2S_2O_8$ + NH_4F	–	–	Decomposes completely
4NaOH (20%) + Br_2 (HBrO, HBr)	20	24	Decomposes up to ~12–19%
	≥100	2	Decomposes up to ~36%
4NaOH (20%) + H_2O_2 (30%)	20	24	Decomposes up to ~47–66%
	100–110	2	Dissolves completely with hydrolytic deposition of metal
4NaOH (20%) + $K_3[Fe(CN)_6]$ (10%)	20	24	Decomposes up to ~14–17%
	≥100	2	Decomposes up to ~63–76%

[a]All the ratios are given in volume fractions

[b]When it is not indicated specially, the character reported is related to the powders with mean particle size of 40–50 μm

[c]The decomposition of HfC_{1-x} starts at 200–220 °C, at 230–250 °C the dissolution is very energetical and decomposition, accompanied by the release of amorphous finely-dispersed C, mainly finishes in 30–40 min; represented by the main reaction equations:

$HfC_{1-x} + 2H_2SO_4 = Hf(SO_4)_2 + (1 - x)CH_4 + 2xH_2$

$HfC_{1-x} + 4H_2SO_4 = Hf(SO_4)_2 + (1 - x)C + 2SO_2 + 4H_2O$

as well as secondary reaction equations:

$C + 2H_2SO_4 = CO_2 + 2H_2O + 2SO_2$

$CH_4 + 4H_2SO_4 = CO_2 + 6H_2O + 4SO_2$

[d]Represented by the following reaction equation:

$HfC_{1-x} + 6HNO_3 = HfO(NO_3)_2 + (1 - x)CO_2 + 2(1 - x)NO + 2(1 + x)NO_2 + 3H_2O$

[e]Represented by the following reaction equation:

$HfC_{1-x} + 2H_3PO_4 + H_2O = HfO(H_2PO_4)_2 + (1 - x)CH_4 + 2xH_2$

[f]Oxalic acid

(continued)

Table 3.24 (continued)

[g]Recommended chemical etching agents for HfC_{1-x}: **(a)** boiling mixture of concentrated H_2SO_4 + $(NH_4)_2S_2O_8$ [337]; **(b)** 10 g $K_3[Fe(CN)_6]$ + 2–10 g KOH/NaOH + 100 ml H_2O (Murakami's reagent for metallographic analysis) [185]; **(c)** 5% Br_2 solution in CH_3OH (methanol) (for the separation of carbide phases from metal-carbide alloys) [185]

[h]Tartaric acid

[i]Citric acid

[j]Ethylenediaminetetraacetic acid (EDTA)

In comparison with other ultra-high temperature materials the summarized data on the chemical behaviour of hafnium monocarbide are given in Addendum.

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Chapter 4
Niobium Carbides

4.1 Structures

Niobium forms with carbon several chemical compounds (*see* also section C–Nb in Table I-2.13): niobium monocarbide NbC_{1-x} with the extremely broad homogeneity range, several modifications of niobium semicarbide, such as ordered low-temperature α-Nb_2C and middle-temperature β-$Nb_{2+x}C$ and disordered high-temperature γ-$Nb_{2\pm x}C$, as well as ε-Nb_3C_{2+x}, ζ-Nb_4C_{3-x}, $Nb_5C_{4\pm x}$, $Nb_6C_{5\pm x}$, $Nb_7C_{6\pm x}$ and $Nb_8C_{7\pm x}$ ordered phases [1–7, 154, 415, 861–864, 930, 931, 958, 1095, 1096]. It should be mentioned that there are no forcible scientific arguments in favour of existence of the both ε- and ζ-phases as well as the data on molecular clusters Nb_8C_{12} are not confirmed sufficiently in the literature [4, 8, 12, 841, 891]. A high-temperature partial variant of niobium – carbon phase diagram is given in Fig. 4.1, and the structural features of the niobium carbides are presented in Table 4.1. The C/Nb radii ratio calculated on the basis of Pauling's atomic size of Nb (0.1456 nm, CN = 12) is 0.530 [7], or 0.56 [63]; the ratio of Nb radii (in nm) in Me/MeC is 0.143/0.157 (10.0% expansion of Nb atoms in carbide) [690]. Under the higher pressures of $\sim$60–600 GPa the structural transformation of NbC_{1-x} from NaCl type (*Fm(–3)m*) to CsCl type (*Pm(–3)m*) structure has been predicted theoretically [385–387, 391, 392, 738, 926]. The relationships between the lattice parameter and composition of NbC_{1-x} on the basis of several sources are given in Fig. 4.2; in the modified forms the following equations for $0 \leq x \leq 0.3$ in NbC_{1-x} earlier proposed by Kempter et al. [68] and later by Senczyk [1030]:

$$a = 0.4471 - 0.00268x - 0.03457x^2, \tag{4.1}$$

and recently by Gusev et al. (for disordered structures, $0 \leq x \leq 0.30$) [73, 359, 962, 1023]:

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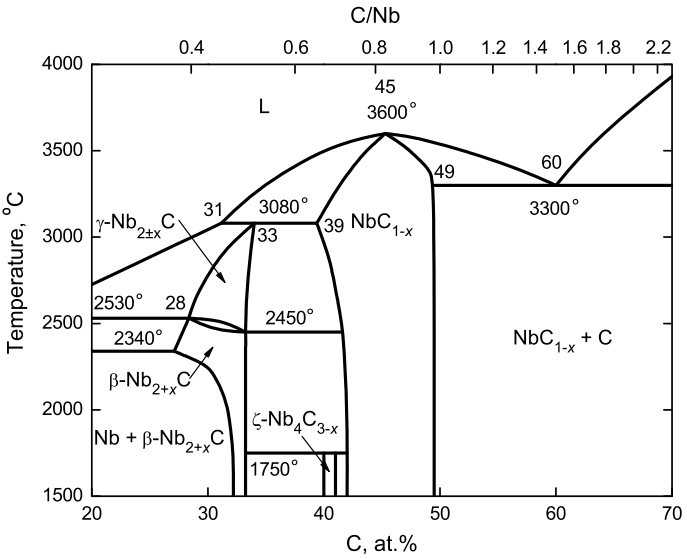

Fig. 4.1 High-temperature partial variant of niobium – carbon equilibrium phase diagram [3–7, 9, 58, 106, 144, 146, 415, 768, 818, 840, 859, 885, 886, 959, 964, 1029, 1034, 1094]

$$a = 0.4467 + 0.00493x - 0.07323x^2, \tag{4.2}$$

$$a = 0.4447 + 0.00378(0.1956 - x)^{1/3}, \tag{4.3}$$

can be recommended to describe it, where a is lattice parameter, nm and x is the value of index in NbC_{1-x} formula; within the homogeneity range of semicarbide $Nb_{2\pm x}C$ phase both the lattice parameters a and c also decline with decreasing carbon content [77, 82, 85]. The minimal Burgers vector of near-stoichiometric NbC_{1-x} (½ <110>) b = 0.316 nm [2]; accordingly to the latter equation [359]:

$$b = 0.3144 + 0.00267(0.1956 - x)^{1/3}. \tag{4.4}$$

The parameters of formation and migration of NbC_{1-x} lattice point defects (vacancies and interstitial atoms) are given in Sect. 4.5 (Table 4.14). At low temperatures for non-stoichiometric monocarbide the slip system (111) <110> was determined, this system is also more preferential for NbC_{1-x} with various x at higher temperatures [10, 11, 87, 292, 297, 298, 690, 698, 1048]; it is necessary to add that for $NbC_{0.87}$ – by Turchin et al. [703] and for $NbC_{0.83}$ ($Nb_6C_{5\pm x}$ phase with non-cubic short- or long-range ordering) – by Morgan and Lewis [296], the coexistence of slip systems (110) <1$\bar{1}$0> and (111) <1$\bar{1}$0> were also determined [713]. The variations of the lattice parameter of NbC_{1-x} with temperature for the phases having different deviations from the stoichiometry are shown in Fig. 4.3.

Table 4.1 Structural properties (crystal structure, density) of niobium carbide phases

Formula	Crystal structure							Density[c], g cm^{-3}	Reference
	System	Type	Space group	Lattice parameters[a], nm			Z[b]		
				a	*b*	*c*			
NbC_{1-x}	Cubic	NaCl	*Fm(–3)m*	0.4430[d]	–	–	4	–	[69, 77]
				0.4431[d]	–	–	4	–	[58]
				0.4432[d]	–	–	4	–	[6, 68]
				0.4432[e]	–	–	4	7.79	[46]
				0.4433	–	–	4	–	[66, 67]
				0.4442	–	–	4	–	[27]
				0.4454[f]	–	–	4	–	[18, 69]
				0.4457	–	–	4	–	[64]
				0.4460	–	–	4	–	[65]
				0.4461	–	–	4	–	[62, 159]
				0.4462	–	–	4	7.84	[29]
				0.4463	–	–	4	–	[24]
				0.4464[g]	–	–	4	–	[69]
				0.4465[h]	–	–	4	7.76	[21, 28]
				0.4467	–	–	4	7.82	[39]
				0.4467[i]	–	–	4	7.78	[37, 38]
				0.4468[j]	–	–	4	–	[36]
				0.4468	–	–	4	7.81	[30]
				0.4469	–	–	4	7.81	[10, 11, 22]
				0.4469[k]	–	–	4	7.75	[41, 69]
				0.4469[l]	–	–	4	7.70	[59]
				0.4470[m]	–	–	4	–	[58, 80]
				0.4470[m]	–	–	4	7.78	[25, 31, 32]
				0.4471	–	–	4	7.80	[33, 34, 42]
				0.4471[n]	–	–	4	7.77	[43]
				0.4471[o]	–	–	4	7.72	[40]
				0.4471[p]	–	–	4	7.79	[6, 68, 77, 388]
				0.4473	–	–	4	7.80	[26]
γ-$Nb_{2\pm x}C$	Hexagonal	NiAs	*P6$_3$/mmc*	0.3100	–	0.5003[q]	1	7.89	[50]
				0.3114	–	0.4985[r]	1	7.85	[51]
				0.3116	–	0.4958[s]	1	7.88	[56, 57]
				0.3117	–	0.4964[s]	1	7.87	[53]
				0.3118	–	0.4955[s]	1	–	[79]
				0.3118	–	0.4958[s]	1	–	[59]
				0.3118	–	0.4961[s]	1	–	[76]
				0.3119	–	0.4953[s]	1	–	[64]
				0.3119	–	0.4959[s]	1	–	[71, 718]
				0.3120	–	0.4957[s,u]	1	–	[77]
				0.3122	–	0.4964[s]	1	7.79	[84]

(continued)

Table 4.1 (continued)

Formula	Crystal structure						Z[b]	Density[c], g cm^{-3}	Reference
	System	Type	Space group	Lattice parameters[a], nm					
				a	*b*	*c*			
				0.3123	–	0.4966[s,u]	1	–	[82]
				0.3124	–	0.4963[s]	1	7.83	[54]
				0.3125	–	0.4752[t]	1	8.17	[52]
				0.3125	–	0.4983[s]	1	–	[17]
				0.3126	–	0.4965[s]	1	7.85	[10, 11, 69]
				0.3126	–	0.4966[s]	1	–	[61]
				0.3126	–	0.4972[s]	1	–	[2, 3, 6]
				0.3127	–	0.4965[s,u]	1	–	[58]
				0.3127	–	0.4972[s,v]	1	7.80	[6, 58, 81]
				0.3128	–	0.4970[s,v]	1	–	[77]
				0.3128	–	0.4973[s]	1	7.80	[55]
				0.3128	–	0.4974[s]	1	–	[62, 63]
				0.3065–0.3135	–	0.4999–0.4909[w]	1	–	[29]
β-$Nb_{2+x}C^x$	Trigonal	ε-Fe_2N	*P(–3)1m* (or *P(–3)m1*)	0.5401	–	0.4955[y]	3	–	[2, 3, 61]
				0.5407	–	0.4968[y]	3	–	[77]
				0.5407	–	0.4974[y]	3	–	[7, 70]
				0.5408	–	0.4965[y]	3	–	[1, 10, 11]
				0.5417	–	0.4972[y]	3	7.80	[47]
α-Nb_2C^x	Orthorhombic	ζ-Fe_2N	*Pbcn*	1.089	1.236	0.4968	–	–	[61, 77, 79]
				1.076	1.254	0.4966	–	–	[78]
		–	*Pnma*	1.0906	0.3096	0.4969	–	–	[2, 3, 6, 60]
		–	*$Pna2_1$* (or *$Pmc2_1$*)	1.092	0.3090	0.4974	–	–	[7, 35, 69]
				1.074–1.086	0.3099–0.3135	0.4926–0.4907	–	–	[29]
	Monoclinic	–	*Pm*	1.076	0.3135	0.4966	4	7.84	[48]
				1.090	0.3095	0.4967	4	–	[229]
				1.091	0.3095	0.4975	4	7.82	[47]
				1.092	0.3090	0.4970	4	–	[1]
				1.094	0.3099	0.4982	4	7.45	[19]
ε-Nb_3C_{2+x}	Hexagonal	–	–	1.1460	–	1.830[r]	–	–	[12, 15]
ζ-Nb_4C_{3-x}	Trigonal	Sn_4P_3	*R(–3)m*	0.3140	–	3.010[z]	3	7.90	[12, 13, 69]
	Cubic	–	*Pm(–3)m*	0.4445	–	–	1	–	[154, 155]
$Nb_6C_{5\pm x}$	Monoclinic	Nb_6C_5	*C2/m*	0.5447[a1]	0.9435[a1]	0.5447[a1]	2	7.77	[4, 20, 150]
				0.5461[a1]	0.9458[a1]	0.5461[a1]	2	7.72	[14, 16, 49]
	Trigonal	–	*$P3_1$* (or *$P3_2$*)	0.5464[a2]	–	1.5422[a2]	3	7.70	[4, 23, 45, 77, 149]
$Nb_8C^{a3}_{7\pm x}$	Cubic	–	*$P4_332$*	0.8974[a4]	–	–	4	–	[154]

[a]When it is not indicated specially, value reported is for near-stoichiometric compositions
[b]Number of formula units per lattice cell
[c]Calculated from XRD or neutron diffraction patterns
[d]Carbon content – 41.0–41.5 at.% (in equilibrium with semicarbide phase)
[e]Carbon content – 43.2 at.%
[f]Carbon content – 45.0 at.%
[g]Carbon content – 47.0 at.%

(continued)

Table 4.1 (continued)

[h]Carbon content – 47.9 at.%
[i]Carbon content – 49.0 at.%
[j]Carbon content – 48.0 at.%
[k]Carbon content – 48.5–49.0 at.%
[l]Carbon content – 46.9 at.%
[m]Carbon content – 49.5–49.7 at.%
[n]Carbon content – 49.2 at.%
[o]Carbon content – 47.7 at.%
[p]Carbon content – 49.7–49.9 at.% (in equilibrium with graphite)
[q]c/a = 1.61
[r]c/a = 1.60
[s]c/a = 1.59
[t]c/a = 1.52
[u]Carbon content – 32.0–32.9 at.%
[v]Carbon content – 33.3 at.%
[w]c/a = 1.63–1.57
[x]Also structured (vacancy ordered) in *P3m1*, *I4$_1$/amd*, *R(–3)m*, *Fd(–3)m* and *P4/mmm* [930]
[y]c/a = 0.92
[z]c/a = 9.59
[a1]Carbon content – ~45.5 at.%; $\alpha = \gamma = 90°$, $\beta = 109.47°$
[a2]Carbon content – 44.1–44.8 at.%
[a3]Also vacancy ordered $Nb_5C_{4\pm x}$ (*P(−1)*, *C2/m*, *I4/m*) and $Nb_7C_{6\pm x}$ (*R(−3)*) are supposed [930]
[a4]Calculated on the basis of density-functional theory (DFT) with generalized gradient approximation of Perdew-Burke-Ernzerhof (GGA-PBE)

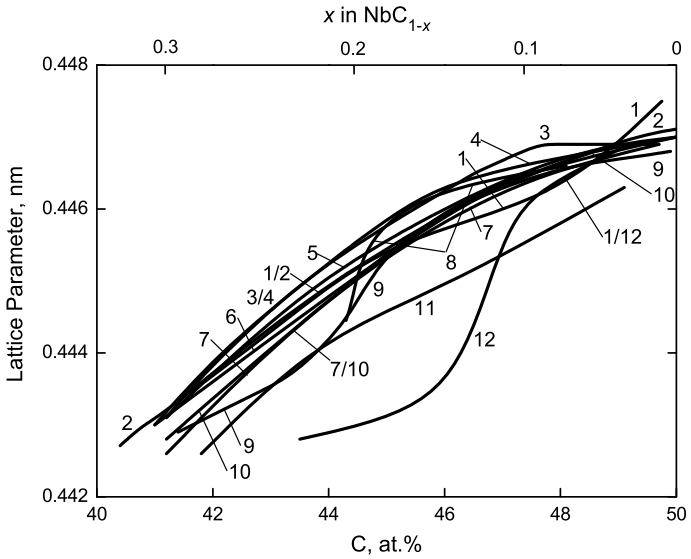

Fig. 4.2 Lattice parameter of NbC_{1-x} as a function of phase composition: 1 – synthesized from the elements [191, 772, 833, 968], 2 – [6, 68, 72], 3 – [82, 85], 4 – [77], 5 – [36, 75], 6 – [69], 7 – polynomial approximation of experimental data [10, 11, 218, 848], 8 and 9 – for ordered and disordered structures, respectively [73, 74, 359, 961–963], 10 – synthesized by different methods [266, 960], 11 – [76], 12 – [83] (1, 3, 5–9 – on the basis of several sources, for single crystal NbC_{1-x} materials *see* Fig. 4.23)

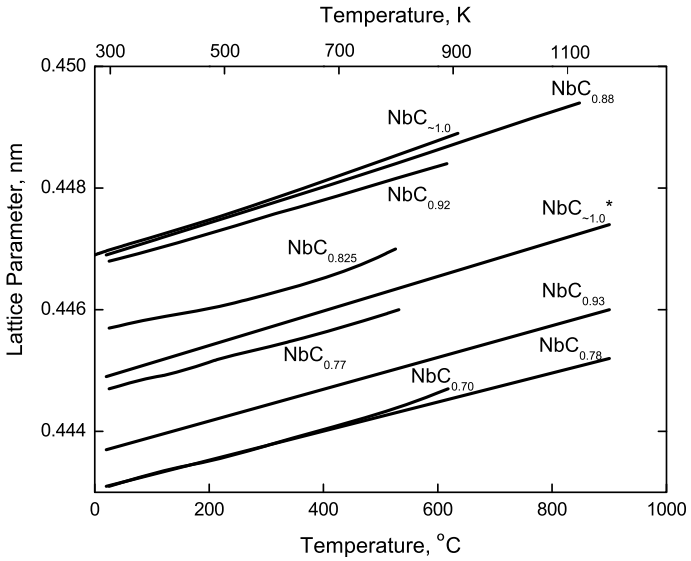

Fig. 4.3 Lattice parameter of NbC_{1-x} as a function of temperature for the phases having different deviations from the stoichiometry ($x \approx 0$ [75]; $x = 0$*, $x = 0.07$ and $x = 0.22$ [699]; $x = 0.08$, $x = 0.175$, $x = 0.23$ and $x = 0.30$ [36]; $x = 0.12$ [82, 86])

In general, the contaminations, such as dissolved in NbC_{1-x} oxygen or/and nitrogen, which are always present in the material, decrease the magnitude of lattice parameter by forming oxycarbide $NbC_{1-x}O_y$, carbonitride $NbC_{1-x}N_z$ and oxycarbonitride $NbC_{1-x}N_zO_y$ phases with various deviations from the stoichiometry (*see also* sections C–N–Nb and C–Nb–O in Table I-2.14) [59, 69, 495, 742, 743].

The strong covalent character of chemical bonding in NbC_{1-x} determines the high value of C–Nb bond energy (enthalpy), which is about 5.89 ± 0.1 eV [870], or 5.85 ± 0.1 eV [871], or 5.43 ± 0.15 eV [130, 872], or 5.39 ± 0.15 eV [873].

Recently, due to the massive progress in nanotechnology, various nanostructures based on niobium carbides, including nanosheets (8–12 nm in sizes) [100], nanocrystalline coatings/films [97–99, 921], nanowires (various types, e.g. 1–3 nm diameter, within the inner space of CNTs) [101, 532], nanorods (2–30 nm diameter, up to 20 μm length) [88, 91, 714], nanotubes (20–60 nm diameter, up to 1–10 μm in length) [92], nanofibres (40–90 nm diameter, up to 1–10 μm in length) [96], whiskers (0.1–3.5 μm diameter, up to 10–100 length/diameter ratios) [89, 90, 884, 936, 969, 970], solid cage shape particles [90] and various (in types) nanoparticles (from 5 to 500 nm) [91, 93–95, 509, 877, 902, 917, 927, 965], have been synthesized, as well as 2D-molecular $Nb_{n+1}C_n$ (or $(Nb,M)_2CT_x$, $Nb_4C_3T_x$, where T are functional groups OH, O, F and others, and M are co-host carbide metals) MXenes [102, 103, 471, 740, 909, 918, 920, 923–925, 928, 957, 1019].

According to Storms et al. [6, 68], the X-ray density of niobium monocarbide NbC_{1-x} phase drops from 7.79 g cm^{-3} for $NbC_{0.99}$ (value extrapolated to $NbC_{1.0}$ phase – 7.796 g cm^{-3}) to a minimum of 7.72 g cm^{-3} for $NbC_{0.79}$ and then rises to

7.73 g cm^{-3} for the composition corresponding to the NbC_{1-x}–$Nb_{2\pm x}C$ phase boundary, while the X-ray density of $Nb_{2\pm x}C$ semicarbide phase in equilibrium state with NbC_{1-x} monocarbide is 7.80 g cm^{-3}. The recommended values for the bulk (or pycnometric) density of pure poreless materials at room temperature are about 7.80–7.85 and 7.75–7.80 g cm^{-3} – for the near-stoichiometric single-phase compositions of niobium semicarbide $Nb_{2\pm x}C$ and monocarbide NbC_{1-x}, respectively [130, 307, 308, 719, 818, 1034].

4.2 Thermal Properties

Niobium monocarbide NbC_{1-x} has one of the highest melting points of all the solid substances available; the incongruent melting point of γ-$Nb_{2\pm x}C$ is about 500° lower. Within the homogeneity range of NbC_{1-x} the melting point of the phase varies (*see* Fig. 4.1); the maximum temperatures pertain to the non-stoichiometric compositions around $\sim NbC_{0.79–0.84}$ [6, 9, 105, 106]. The general thermodynamic properties of near-stoichiometric niobium carbide phases are summarized in Tables 4.2 and 4.3. For the molar heat capacity of near-stoichiometric niobium monocarbide NbC_{1-x} $c_p = f(T, \text{K})$, J mol^{-1} K^{-1}, the following relationships were recommended in the literature:

in the range of temperatures from 50 to 1600 K [80]

$$c_p = 56.07\exp(-115.40/T), \tag{4.5}$$

in the range of temperatures from 300 to 670 K [124, 127, 128]

$$c_p = 45.94 + \left(5.78 \times 10^{-3}\right)T - \left(9.71 \times 10^{5}\right)T^{-2}, \tag{4.6}$$

in the range of temperatures from 300 to 1800 K [125, 127, 128]

$$c_p = 45.15 + \left(7.22 \times 10^{-3}\right)T - \left(9.00 \times 10^{5}\right)T^{-2}, \tag{4.7}$$

in the range of temperatures from 500 to 2400 K [126–128]

$$c_p = 47.03 + \left(5.23 \times 10^{-3}\right)T - \left(10.63 \times 10^{5}\right)T^{-2}, \tag{4.8}$$

in the range of temperatures from 600 to 2500 K [80]

$$c_p = 43.06 + \left(7.76 \times 10^{-3}\right)T. \tag{4.9}$$

in the range of temperatures from 670 to 1800 K [124, 127, 128]

$$c_p = 47.20 + \left(4.94 \times 10^{-3}\right)T - \left(10.46 \times 10^{5}\right)T^{-2}, \tag{4.10}$$

Table 4.2 General thermodynamic properties of near-stoichiometric niobium monocarbide NbC_{1-x}

Characteristics	Symbol	Unit	Value	Reference
Standard heat of formation (at 298.15 K)[a]	$-\Delta H°_{298}$	kJ mol^{-1}	125.6	[117]
			129.7 ± 2.5[b]	[109]
			132.8 ± 3.4[c]	[108, 140]
			133.1 ± 3.8[d]	[141]
			134.3	[113]
			134.7	[82, 107]
			138.2	[142]
			139.7[e]	[1, 721]
			140.6 ± 10.0[f,g]	[4–7, 62–64, 108, 111, 118, 753, 1021–1023]
			142.1 ± 2.7[g]	[109]
			145.6	[112]
Standard molar entropy[h] (at 298.15 K and 100 kPa)	$S°_{298}$	J mol^{-1} K^{-1}	34.69[e]	[1, 6]
			34.89	[141]
			34.97[e]	[127–129]
			35.06	[122]
			35.28	[140]
			35.40 ± 0.20	[114, 124]
			36.40	[5, 132]
			37.07[i]	[125]
			37.25 ± 2.9	[62, 63, 67, 82, 118]
			37.66	[125, 170]
Molar enthalpy difference	$H_{298} - H_0$	kJ mol^{-1}	5.422[e]	[127, 128]
Standard molar heat capacity[j] (at 298.15 K and 100 kPa)	$c°_{p,298}$	J mol^{-1} K^{-1}	33.24[k]	[74, 143]
			33.35[l]	[6, 122]
			33.70[m]	[74, 143]
			33.78[n]	[74, 143]
			34.38[o]	[74, 143]
			34.70[p]	[74, 143]
			34.73[q]	[74, 143]
			36.23[e]	[127–129]
			36.62	[126]
			36.78[e]	[1, 2, 6, 122]
			36.82[r]	[7, 67, 80, 84, 114, 124]
			37.18	[5, 125, 132]
			37.35	[62, 63, 82]
Specific heat capacity (at 298.15 K)	c	J kg^{-1} K^{-1}	354.4	[5, 132]
			355.9	[62, 116]

(continued)

Table 4.2 (continued)

Characteristics	Symbol	Unit	Value	Reference
Molar enthalpy (heat) of melting (at the melting point)	ΔH_m	kJ mol^{-1}	92	[5, 132]
Specific enthalpy (heat) of melting (at the melting point)		kJ kg^{-1}	868	[5, 132]
Molar enthalpy (heat) of vaporization (dissociation)[s]	$\Delta \hat{H}_v$	kJ mol^{-1}	630–790[t]	[67, 160]
Melting point	T_m	K (°C)	3660 ± 20 (3390 ± 20)	[1]
			3750 (3480)	[62, 63, 110]
			3755 ± 50 (3485 ± 50)	[146]
			3760 (3490)	[119]
			3770 ± 100 (3500 ± 100)[u]	[64, 104, 117, 159, 295, 338, 720, 752]
			3775 ± 75 (3505 ± 75)[u]	[58]
			3800 ± 35 (3530 ± 35)	[120]
			3870 ± 50 (3600 ± 50)[o,v]	[2, 5–7, 71, 84, 121, 123, 721, 808, 905]
			3880 (3610)[w]	[3, 130]
			3885 ± 25 (3615 ± 25)[x,y]	[67, 137, 144, 145, 552, 751]
Boiling point	T_b	K (°C)	4570 (4300)	[64, 117, 130]
			4770 (4500)	[62, 63, 67, 118, 138]
			5275 (5000)	[751]

[a]Enthalpy (heat) of complete dissociation (atomization) from solid state at 298.15 K ($-\Delta_{at}H°_{298}$, kJ mol^{-1}): 1537 ± 9 [1], 1649 [748], 1598 [749], 1674 [750], 1580 ± 10 [1023]
[b]Carbon content – 47.7 at.%
[c]Carbon content – 48.6 at.%
[d]Carbon content – 49.7 at.%
[e]Carbon content – 49.5 at.%
[f]On the basis of several sources
[g]Data extrapolated to the stoichiometric composition of monocarbide
[h]Molar entropy $S°_T$ (at 1200 K), J mol^{-1} K^{-1}, for NbC_{1-x} phases with different deviations from the stoichiometry: 101.61 (x = 0), 99.96 (x = 0.05), 96.17 (x = 0.20), 95.29 (x = 0.25) and 94.61 (x = 0.30) [128]
[i]Carbon content – 42.8 at.%
[j]Molar heat capacity $c_{p,T}$ (at 1200 K), J mol^{-1} K^{-1}, for NbC_{1-x} phases with different deviations from the stoichiometry: 55.44 (x = 0), 53.89 (x = 0.05), 50.34 (x = 0.20), 49.52 (x = 0.25) and 48.88 (x = 0.30) [128]
[k]Disordered structure with carbon content – 44.8 at.%
[l]Carbon content – 41.2 at.%
[m]Ordered structure with carbon content – 44.8 at.%
[n]Disordered structure with carbon content – 45.4 at.%
[o]Ordered structure with carbon content – 45.4 at.%
[p]Disordered structure with carbon content – 46.8 at.%

(continued)

Table 4.2 (continued)

[q]Ordered structure with carbon content – 46.8 at.%
[r]Carbon content – 49.9 at.%
[s]Enthalpy (heat) of sublimation of metallic Nb at 298.15 K, $\Delta_{sMe}H°_{298}$ = 723.1 kJ mol^{-1} [1024]
[t]For the process of NbC_{1-x} (crystal) = NbC_{1-x-y} (crystal) + yC (gas) at 2800 K (2530 °C)
[u]Carbon content – ~46 at.%
[v]Carbon content – ~50 at.%
[w]For $NbC_{0.78\pm0.03}$ composition
[x]Carbon content – 44.0 ± 1.0 at.% [67, 137, 145]
[y]Carbon content – 45.0 ± 1.0 at.% [144]

Table 4.3 General thermodynamic properties of near-stoichiometric niobium semicarbide $Nb_{2\pm x}C$

Characteristics	Symbol	Unit	Value	Reference
Standard heat of formation (at 298.15 K)[a]	$-\Delta H°_{298}$	kJ mol^{-1}	156.5[b]	[10, 82, 115]
			180.3[c]	[111]
			188.3	[1]
			190.0	[67, 114]
			190.2	[118]
			195.0 ± 5.0	[6, 7, 84, 108, 111, 112, 140, 753]
Standard molar entropy (at 298.15 K and 100 kPa)	$S°_{298}$	J mol^{-1} K^{-1}	59.80	[63, 82]
			63.50	[132]
			63.80 ± 0.24	[140]
			64.01 ± 0.42	[67, 114, 142]
			64.10	[1, 6, 7, 122]
			73.14	[125]
			73.20	[118]
Standard molar heat capacity (at 298.15 K and 100 kPa)	$c°_{p,298}$	J mol^{-1} K^{-1}	60.52	[125]
			60.67	[67]
			60.72	[62, 63]
			63.51	[1, 6, 7, 114, 122, 125]
			63.56	[2]
Specific heat capacity (at 298.15 K)	c	J kg^{-1} K^{-1}	316.0	[62, 116]
Melting point	T_m	K (°C)	2800 ± 20	[144, 147, 148]
			(2530 ± 20)[d]	
			3200 (2930)	[116]
			3305 ± 20	[7, 67, 137, 145]
			(3035 ± 20)[e]	
			3350 ± 35	[2, 6, 58, 71, 84, 130, 146, 751]
			(3080 ± 35)[f]	
			3360 ± 50	[1, 58, 114, 338]
			(3090 ± 50)[f]	

[a]Enthalpy of complete dissociation (atomization) from solid state at 298.15 K ($-\Delta_{at}H°_{298}$): 2348 ± 14 kJ mol^{-1} [1]
[b]Carbon content – 30.1 at.%
[c]Carbon content – 32.8 at.%
[d]Carbon content – 29.0 ± 2.0 at.%
[e]Carbon content – 34.5 at.%
[f]Carbon content – 33.3 at.%

in the range of temperatures from 1000 to 3000 K [127, 128, 133]

$$c_p = 46.43 + \left(5.87 \times 10^{-3}\right)T - \left(9.01 \times 10^{5}\right)T^{-2}, \tag{4.11}$$

in the range of temperatures from 1600 to 2400 K [127, 128, 134, 135]

$$c_p = 47.18 + \left(4.99 \times 10^{-3}\right)T. \tag{4.12}$$

For the specific heat capacity of near-stoichiometric niobium monocarbide NbC_{1-x} $c = f(T, \text{K})$, J kg^{-1} K^{-1}, the relationship

$$c = 348.9 + \left(89.54 \times 10^{-3}\right)T + \left(326.4 \times 10^{6}\right)T^{-2} \tag{4.13}$$

is recommended for the range of temperatures from 700 to 3200 K [127, 128, 136]. The variations of molar heat capacity with temperature for near-stoichiometric niobium carbides $Nb_{2\pm x}C$ and NbC_{1-x} are demonstrated in Fig. 4.4 on the basis of

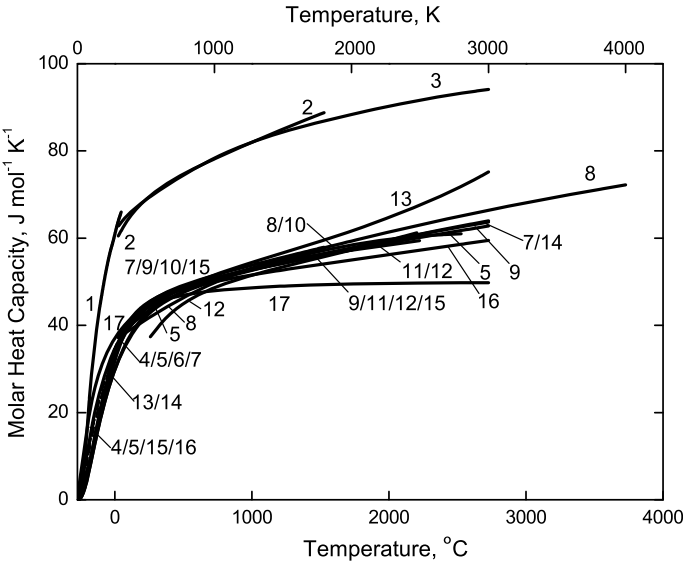

Fig. 4.4 Variation of molar heat capacity (1–14 – at constant pressure, c_p, and 15–17 – at constant volume, c_v) with temperature for near-stoichiometric niobium carbides $Nb_{2\pm x}C$ (1 – [122], 2 – summarized [62], 3 – [125]) and NbC_{1-x} (4 – [122], 5 – [80], 6 – [143], 7 – [151–153], 8 – [5, 132], 9 – [6], 10 – [114], 11 – [139], 12 – $NbC_{0.93}$ [181, 338]; theoretically calculated: 13 – Slater's approximation and 14 – Dugdale-MacDonald's approximation on the basis of Debye-Grüneisen model [152], 15 – *ab initio* calculated, based on the mean values between the corresponding generalized gradient approximation (GGA) and local density approximation (LDA) estimations [409], 16 – using the GGA with the Perdew-Wang parameterization (PW91) [356], 17 – using projector augmented-wave (PAW) method within the GGA of the Perdew-Burke-Ernzerhof scheme for ionic solids (PBEsol) [796]; 5, 7–10 – on the basis of several sources)

several sources. For the molar heat capacity of monocarbide NbC_{1-x} phases with various deviations from the stoichiometry the following relationships were recommended in literature:

for $NbC_{0.75}$ (in the ranges of temperatures from 300 to 1800 K) [125, 127, 128]

$$c_p = 37.45 + \left(9.43 \times 10^{-3}\right)T - \left(5.27 \times 10^{5}\right)T^{-2}, \tag{4.14}$$

for $NbC_{0.75}$ (in the ranges of temperatures from 1440 to 2500 K) [128, 139, 156]

$$c_p = 42.22 + \left(4.76 \times 10^{-3}\right)T + \left(82.76 \times 10^{10}\right)T^{-2}\exp(-22910/T); \tag{4.15}$$

for $NbC_{0.86}$ (in the ranges of temperatures from 1270 to 2600 K) [128, 139, 156]

$$c_p = 47.28 + \left(3.48 \times 10^{-3}\right)T + \left(31.92 \times 10^{9}\right)T^{-2}\exp(-16300/T), \tag{4.16}$$

for $NbC_{0.87}$ (in the ranges of temperatures from 300 to 1800 K) [125, 127, 128]

$$c_p = 40.58 + \left(8.33 \times 10^{-3}\right)T - \left(6.32 \times 10^{5}\right)T^{-2}, \tag{4.17}$$

for $NbC_{0.91}$ (in the ranges of temperatures from 1230 to 2515 K) [128, 139, 156]

$$c_p = 48.14 + \left(3.85 \times 10^{-3}\right)T + \left(6.40 \times 10^{9}\right)T^{-2}\exp(-14260/T), \tag{4.18}$$

for $NbC_{0.97}$ (in the ranges of temperatures from 1300 to 2800 K) [127, 128, 131]

$$c_p = 46.50 + \left(5.38 \times 10^{-3}\right)T, \tag{4.19}$$

for $NbC_{0.99}$ (in the ranges of temperatures from 1300 to 2500 K) [128, 139, 156]

$$c_p = 48.46 + \left(4.37 \times 10^{-3}\right)T \tag{4.20}$$

and for NbC_{1-x} (in the ranges of temperatures from 1200 to 2200 K) in general [128, 157]

$$c_p = \left(1 - 0.83x + 1.07x^2\right)\left[48.50 + \left(4.436 \times 10^{-3}\right)T\right], \tag{4.21}$$

or

$$c_p = 51.72 + \left(3.10 \times 10^{-3}\right)T - 32.74x + 36.29x^2, \tag{4.22}$$

where T is temperature, K, and x is the value of index in NbC_{1-x} formula. For the molar heat capacity of niobium semicarbide $Nb_{2\pm x}C$ phase ($x = 0$) Geld and Kusenko [125] proposed the relationship as follows

$$c_p = 33.22 + \left(6.28 \times 10^{-3}\right)T - \left(4.29 \times 10^{5}\right)T^{-2}, \tag{4.23}$$

It should be emphasized specially that all the thermal properties of niobium carbides are very sensitive to the deviation from the stoichiometry in the phases: for the standard heat of formation ΔH°_{298}, kJ mol^{-1}, molar enthalpy difference $H^\circ_T - H^\circ_{298}$, J mol^{-1}, and molar entropy S°_T, J mol^{-1} K^{-1}, the following concentration and temperature-concentration dependencies within the homogeneity range of NbC_{1-x} were obtained by the calculations based on experimental data [4, 5, 10, 11, 107, 111, 113, 128, 157, 158, 755, 1021–1023]

$$\Delta H^\circ_{298} = 58.58x - 134.69, \tag{4.24}$$

$$\Delta H^\circ_{298} = 56.07x - 134.31, \tag{4.25}$$

$$\Delta H^\circ_{298} = 27.62 - 297.0(1-x) + 128.7(1-x)^2 \pm 10.0, \tag{4.26}$$

$$H^\circ_T - H^\circ_{298} = (1 - 0.83x + 1.07x^2)\,[48.50T + (2.218 \times 10^{-3})T^2 - 19600], \tag{4.27}$$

$$H^\circ_T - H^\circ_{298} = 51.72T + (1.550 \times 10^{-3})T^2 + 2845x - 32.74xT + 36.29x^2T - 23090, \tag{4.28}$$

$$S^\circ_T = 119.11\lg T + (3.100 \times 10^{-3})T - 80.01x\lg T + 83.58x^2\lg T + 211.37x - 218.46x^2 - 268.87, \tag{4.29}$$

where T is temperature, K, and x is the value of index in NbC_{1-x} formula (Eqs. (4.27)–(4.29) were recommended for the range of temperatures from 1200 to 2200 K). The thermodynamic functions of quasi-stoichiometric niobium monocarbide $NbC_{0.98}$ are tabulated by Turchanin et al. [128, 129] and Chase [754] in the range of 0–3000 K, thermodynamic functions of stoichiometric $NbC_{\sim 1.0}$ – by Schick [132] in the range of 0–6000 K, thermodynamic functions of $NbC_{0.702}$, $NbC_{0.825}$ and $NbC_{\sim 1.0}$ – by Barin [114] in the range of 298.15–1800 K, and thermodynamic functions of $NbC_{0.75}$, $NbC_{0.87}$ and $NbC_{0.98}$ – by Storms [6] and Toth [7] in the range of 298.15–3000 K. For stoichiometric niobium semicarbide the thermodynamic functions are tabulated by Barin [114] in the range of 298.15–1800 K, by Schick [132] in the range of 0–3363 K, and by Storms [6], Toth [7] and Turchanin A and Turchanin M [128] in the range of 298.15–3000 K (for the composition of niobium semicarbide $Nb_{2\pm x}C$ with $x = 0$).

During the vaporization processes from the surface of niobium carbides in a vacuum at high and ultra-high temperatures, the composition of the carbide phase (C/Nb ratio) can change significantly. The resulting composition gradient cannot be accurately evaluated, but it was found to become worse as the temperature was increased [6]. The following equations were recommended for niobium (P_{Nb}, Pa) and carbon (P_C, Pa) partial pressures over near-stoichiometric monocarbide phase [5, 6, 10, 11, 160] for the range of about 2500–3000 K (2200–2700 °C):

$$\lg P_{\mathrm{Nb}} = -\left(4.542 \times 10^4\right)/T + 13.159, \tag{4.30}$$

$$\lg P_{\mathrm{C}} = -\left(3.276 \times 10^4\right)/T + 10.302, \tag{4.31}$$

the estimation of niobium partial pressures (P_{Nb}, Pa) in the gaseous phase in the Nb–C system carried out by Kulikov [751] led to such relationships as

for the conditions of $Nb_{2\pm x}C$–NbC_{1-x} phases equilibrium:

at 1000–2740 K (730–2470 °C)

$$\lg P_{\mathrm{Nb}} = -\left(4.056 \times 10^4\right)/T + 13.273 \tag{4.32}$$

at 2740–3750 K (2470–3480 °C)

$$\lg P_{\mathrm{Nb}} = -\left(3.950 \times 10^4\right)/T + 12.888, \tag{4.33}$$

for the conditions of congruent vaporization of quasi-stoichiometric NbC_{1-x}:

at 1000–2740 K (730–2470 °C)

$$\lg P_{\mathrm{Nb}} = -\left(4.115 \times 10^4\right)/T + 13.028, \tag{4.34}$$

at 2740–3750 K (2470–3480 °C)

$$\lg P_{\mathrm{Nb}} = -\left(4.062 \times 10^4\right)/T + 12.840, \tag{4.35}$$

at 3750–5000 K (3480–4730 °C)

$$\lg P_{\mathrm{Nb}} = -\left(3.538 \times 10^4\right)/T + 11.444, \tag{4.36}$$

for the conditions of NbC_{1-x}–C phases equilibrium:

at 1000–2740 K (730–2470 °C)

$$\lg P_{\mathrm{Nb}} = -\left(4.479 \times 10^4\right)/T + 12.836, \tag{4.37}$$

at 2740–3750 K (2470–3480 °C)

$$\lg P_{\mathrm{Nb}} = -\left(4.331 \times 10^4\right)/T + 12.298, \tag{4.38}$$

at 3750–5000 K (3480–4730 °C)

$$\lg P_{\mathrm{Nb}} = -\left(3.726 \times 10^4\right)/T + 10.685, \tag{4.39}$$

Table 4.4 Parameters of the gaseous phase in the Nb–C system in the conditions of $Nb_{2\pm x}C$–NbC_{1-x} phases equilibrium, congruent vaporization of quasi-stoichiometric NbC_{1-x} and NbC_{1-x}–C phases equilibrium calculated on the basis of thermodynamic data [751]

Parameters	Temperature, K (°C)								
	1000 (730)	1500 (1230)	2000 (1730)	2500 (2230)	2740 (2470)	3000 (2730)	3750 (3480)	4000 (3730)	5000 (4730)
				$Nb_{2\pm x}C$–NbC_{1-x} phases equilibrium					
lgP_{Nb}, Pa	−27.282	−13.720	−6.962	−2.933	−1.523	−0.271	2.363	–	–
lgP^{a}_{Σ}, Pa	−27.273	−13.698	−6.929	−2.889	−1.474	−0.210	2.430	–	–
C/Nb[b]	0.021	0.052	0.080	0.106	0.119	0.152	0.172	–	–
Contents, vol.%:									
C	2.08	4.91	7.39	9.58	10.53	13.00	13.86	–	–
C_2	–	–	–	0.02	0.05	0.10	0.46	–	–
Nb	97.92	95.09	92.61	90.40	89.42	86.90	85.68	–	–
				Congruent dissociation of quasi-stoichiometric NbC_{1-x}					
lgP_{Nb}, Pa	−28.118	−14.363	−7.510	−3.416	−1.979	−0.698	2.017	2.636	4.368
lgP^{a}_{Σ}, Pa	−27.818	−14.062	−7.210	−3.117	−1.683	−0.406	2.291	2.908	4.638
Contents, vol.%:									
C	50.00	50.00	49.92	49.40	48.61	47.48	42.32	41.20	40.19
C_2	–	–	0.04	0.32	0.65	1.13	3.14	3.76	4.75
C_3	–	–	0.01	0.08	0.19	0.40	1.39	1.60	1.33
Nb	50.00	50.00	50.03	50.20	50.55	50.99	53.15	53.44	53.72
				NbC_{1-x}–C phases equilibrium					
lgP_{Nb}, Pa	−31.952	−16.951	−9.502	−5.056	−3.504	−2.128	0.757	1.378	3.233
lgP^{a}_{Σ}, Pa	−24.287	−11.719	−5.224	−1.156	0.288	1.595	4.358	5.034	6.393
lg(C/Nb)[c]	7.665	4.950	4.568	4.292	4.210	4.154	4.070	4.140	3.627
Contents, vol.%:									
Nb	–	–	0.005	0.010	0.016	0.019	0.025	0.023	0.069

[a]Total gas pressure
[b]Atomic ratio in gaseous phase
[c]Logarithm of atomic ratio in gaseous phase

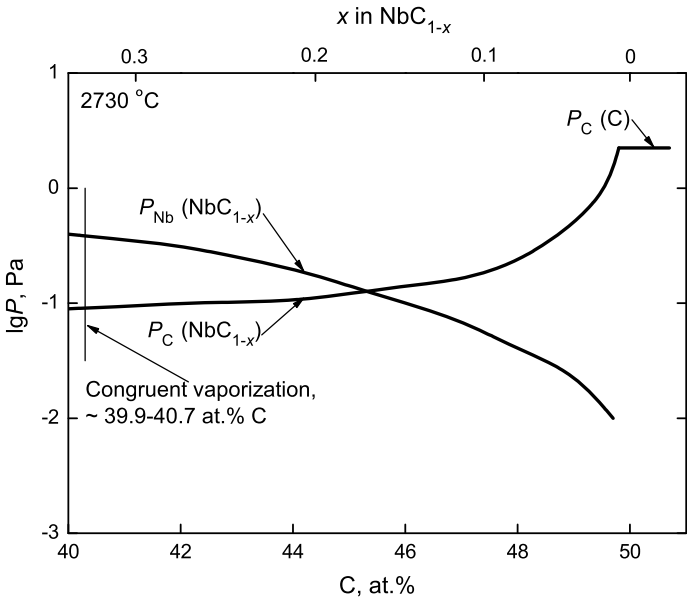

Fig. 4.5 Partial pressures of niobium P_{Nb} (NbC_{1-x}) and carbon P_C (NbC_{1-x}) over niobium monocarbide NbC_{1-x} phase, and vapour pressure of carbon over pure graphite P_C (C) as functions of monocarbide composition at 3000 K (2730 °C) calculated by Kaufman [5, 7, 10, 11, 161]; an approximate congruently vaporizing composition is marked

where T is temperature, K (*see* Table 4.4). The partial pressures of niobium and carbon over the monocarbide phase NbC_{1-x} as functions of composition at the fixed temperature 3000 K (2730 °C) are shown in Fig. 4.5 in comparison with the equilibrium pressure of carbon over graphite surface. The presence in the vapour over NbC_{1-x} phase of stable gaseous carbide molecules NbC and NbC_2 as well as carbon clusters C_2, C_3, …, C_7 becomes noticeable only at temperatures higher than 2700 K (2400 °C) [871, 967]. The vapour phase above carbon-saturated niobium monocarbide NbC_{1-x} in equilibrium with graphite contains atoms of niobium and molecules of carbon plus molecules of NbC_2, the relative concentration of which is determined at 2500 K (2230 °C) by the ratio $Nb/NbC_2 \approx 40$ [147]. The equilibrium pressure should be lower than the pressure observed upon vaporization from an open surface into a vacuum (Langmuir mode). According to Nikolskaya et al. [113, 162, 164, 167] the vapour pressures of niobium (P_{Nb}, Pa) and carbon (P_C, Pa) in Langmuir vaporization of niobium monocarbide can be described by the equations

$$\lg P_{Nb} = 15.16 - \left(4.148 \times 10^4\right)/T - (2.3 \times 10^{-4})T + \lg x, \tag{4.40}$$

$$\lg P_C = 11.31 - \left(4.041 \times 10^4\right)/T + (1.1 \times 10^{-4})T + \lg[(1-x)/x], \tag{4.41}$$

where x is the value of index in NbC_{1-x} formula and T is temperature, K. The vaporization of NbC_{1-x} in vacuum occurs in a more complex manner than the other transition metal refractory carbides. Fesenko and Bolgar [163, 843] proposed for the evaporation rate of $NbC_{0.77}$ G, g cm^{-2} s^{-1}, as a function of temperature T, K, at 2770–3170 K (2500–2900 °C) the following equation:

$$\lg G = 9.0 - (4.33 \times 10^4)/T, \tag{4.42}$$

the linear evaporation rates of $NbC_{0.77}$ are 3.8×10^{-10} sm s^{-1} at 2230 °C and 6.5×10^{-7} sm s^{-1} at 2830 °C [727]; in vacuum ($\sim 10^{-3}$ Pa) Gusev [966] obtained for the mass evaporation rates of $NbC_{0.92}$ (content O – 0.27%) at 2150, 2300 and 2400 °C the values of 4.3×10^{-8}, 7.8×10^{-8} and 1.1×10^{-7} g cm^{-2} s^{-1}, respectively, and for $NbC_{0.98}$ (contents: non-combined C – 0.11%, O – 0.09%) at 1800, 2000 and 2300 °C the values of 6.0×10^{-9}, 1.7×10^{-8} and 7.4×10^{-8} g cm^{-2} s^{-1}, respectively, and established for the general evaporation rate G, g cm^{-2} s^{-1}, at 1900–2700 K (1600–2400 °C) the relationship

$$\lg G = -(2.58 \pm 0.01) - (1.17 \times 10^4)/T. \tag{4.43}$$

In the process of evaporation the lattice parameter on the surface of niobium monocarbide NbC_{1-x} phases gradually decrease, reaching certain constant value for a given temperature [966]. The congruently vaporizing composition of niobium monocarbide shifts significantly in the direction of enrichment of the carbide phase with metal as the temperature increases [10, 11, 113, 162–169, 843]: NbC_{1-x} vaporizes congruently from the composition $NbC_{0.83}$ at 2450 K (2180 °C) [162] to compositions $NbC_{0.77}$ at 2770 K (2500 °C) [163], $NbC_{0.748}$ at 2910 K (2640 °C) [6, 160] and $NbC_{0.70}$ at 3430 K (3160 °C) [162] – it depends on temperature in the range of 2300–3400 K (2030–3130 °C) according to the following relationship [10, 11, 113, 162]

$$\lg x = 0.1388 - \left(2.222 \times 10^3\right)/T, \tag{4.44}$$

where x is the value of index in NbC_{1-x} formula and T is temperature, K. The integral rate of congruent vaporization V, g cm^{-2} s^{-1} of niobium monocarbide NbC_{1-x} at 2480–3430 K (2210–3160 °C) depends on temperature as follows [162, 167]

$$\lg V = 9.406 - \left(4.127 \times 10^4\right)/T, \tag{4.45}$$

where T is temperature, K. Niobium semicarbide γ-$Nb_{2\pm x}C$, when it vaporizes from an open surface into a vacuum preferentially loses the metal; the equilibrium vapour pressure of niobium over the lower carbide is considerably higher than over niobium monocarbide [10, 11].

The values of general thermodynamic properties, vapour pressures and mass/linear vaporization rates for niobium carbides are given in Addendum in comparison with other ultra-high temperature materials in the wide ranges of temperatures.

At room temperature thermal conductivity of near-stoichiometric niobium monocarbide NbC_{1-x}, affected noticeably by porosity, is about 15–40 W m^{-1} K^{-1} (with thermal diffusivity of around 0.13 cm^2 s^{-1}) [1, 2, 7, 10, 11, 62–65, 80, 104, 116, 117, 159, 170–179, 250, 338, 339, 719, 722, 723, 735, 736, 905], for near-stoichiometric niobium semicarbide α-Nb_2C the corresponding value is about 12 W m^{-1} K^{-1} [1]. Marmer et al. [170] revealed for NbC_{1-x} materials in the range of 1000–2400 °C the increase of thermal conductivity from 8–13 to 11–17 W m^{-1} K^{-1} with 7% increase of porosity, it should be mentioned especially that the same researchers also found out the similar effect of porosity on TiC_{1-x} and ZrC_{1-x} materials. Within the homogeneity range, as a consequence of conduction electrons scattering on the carbon sublattice vacancies and thermal lattice vibrations, the thermal conductivity of NbC_{1-x} declines with increasing carbon deficit in the phase [10, 11, 171, 172] that is shown in Fig. 4.6. The variations of thermal conductivity with temperature for near- and non-stoichiometric NbC_{1-x} phases on the basis of several sources are given in Figs. 4.7 and 4.8.

At room temperature the mean coefficients of linear thermal expansion of near-stoichiometric semicarbide α-Nb_2C and monocarbide NbC_{1-x} phases are (6.7–8.1) × 10^{-6} K^{-1} and (5.7–6.9) × 10^{-6} K^{-1}, respectively [10, 11, 82, 84, 177,

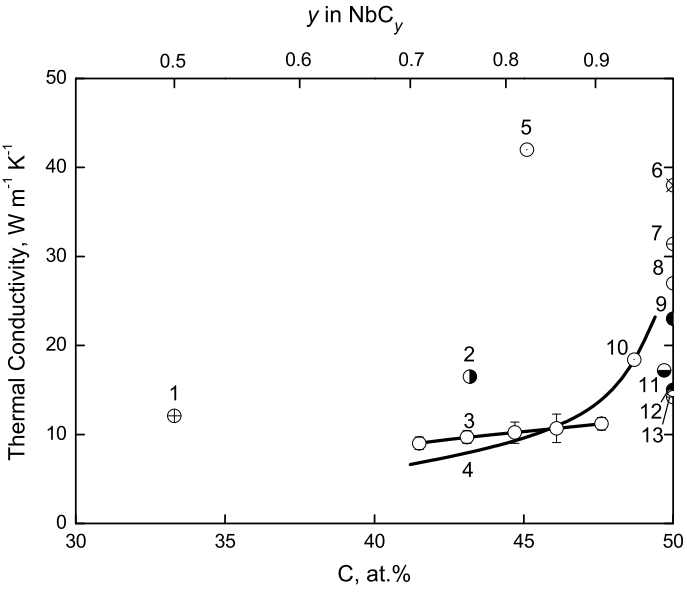

Fig. 4.6 Thermal conductivity at room temperature of niobium semicarbide α-Nb_2C (1 – calculated on the basis of Goryachev's method [1]) and monocarbide NbC_{1-x} (within the homogeneity range: 2 – single crystal [174, 175]; 3 – sintered in vacuum, 5–15% porosity [10, 11, 171]; 4 – hot-pressed at 2730 °C and subsequently annealed at 1850–1950 °C in vacuum, corrected to the poreless state [172]; 5 – single crystal [2, 179]; 6 – [170]; 7 – [80]; 8 – [178]; 9 – [65]; 10 – hot-pressed (measured at 0 °C) [176]; 11 – [173]; 12 – calculated on the basis of Goryachev's method [1]; 13 – [62–64, 104, 116, 117, 159, 177]) as a function of carbides composition

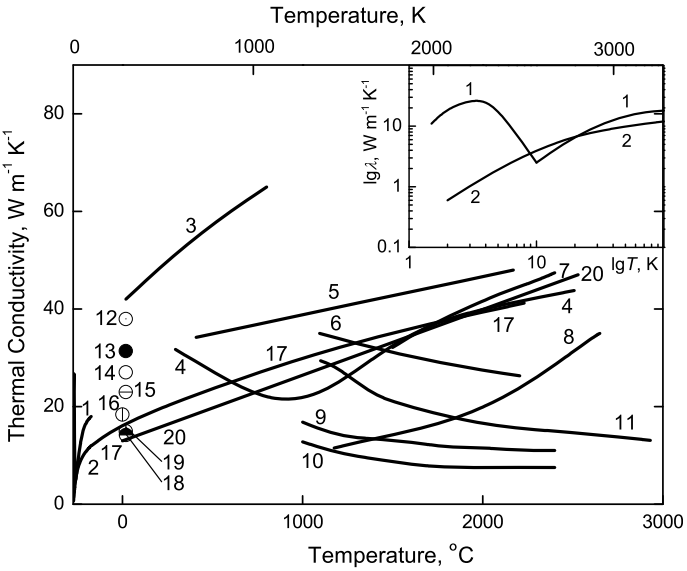

Fig. 4.7 Variation of thermal conductivity λ with temperature for near-stoichiometric (when it is not indicated specially) niobium monocarbide NbC_{1-x} on the basis of several sources: 1 – single crystal, 49.0 at.% C (superconductor with transition at 9.8 K) [182, 183, 194]; 2 – single crystal, 43.2 at.% C [182, 183]; 3 – single crystal, 45.1 at.% C [179]; 4 – heat-soaked at 2480 °C [104, 181]; 5 – sintered in vacuum, ~50 at.% C, corrected to the poreless state [226, 227]; 6 – as-hot-pressed [104, 181]; 7 – vapour deposited, 43.5 at.% C [10, 11, 250]; 8 – hot-pressed [10, 11]; 9 – 27% porosity [170]; 10 – 20% porosity [170]; 11 – hot-pressed, 13–18% porosity, 49.2 at.% C [5, 180, 846]; 12 – [170]; 13 – [80]; 14 – [178, 735, 736]; 15 – [65]; 16 – hot-pressed, 48.7 at.% C [176]; 17 – hot-pressed, 3% porosity, 49.7 at.% C [173]; 18 – [7, 62–64, 104, 116, 117, 159, 177, 719, 722, 723]; 19 – calculated on the basis of Goryachev's method [1]; 20 – summarized on the basis of various sources [173, 250, 338, 339] (*Inset* – at ultra-low and low temperatures in $\lg\lambda - \lg T$, K scale [182, 183, 194, 711, 852])

178, 195, 1031]; along the main crystallographic directions *a* and *c* of Nb_2C, according to Samsonov et al. [10, 67]: $\alpha_{a,m} = 7.0 \times 10^{-6}$ K^{-1} and $\alpha_{c,m} = 8.7 \times 10^{-6}$ K^{-1}. The experimental data collected from the various measurements of thermal expansion of niobium monocarbide NbC_{1-x} are listed in Table 4.5. The approximation function for the temperature dependence of relative thermal linear expansion $\Delta l/l_0 = f(T, \text{K})$, %, of near-stoichiometric NbC_{1-x} was recommended by Touloukian et al. [1031] on the basis of experimental data for pure samples from several sources (accuracy within ±5%) for 293–3000 K (20–2730 °C):

$$\Delta l/l_0 = -0.158 + (5.138 \times 10^{-4})T + (1.023 \times 10^{-7})T^2 + (5.718 \times 10^{-13})T^3, \quad (4.46)$$

where *T* is temperature, K. On the basis of high-temperature X-ray measurements Houska [184] proposed for the average coefficient of linear thermal expansion α_m, K^{-1}, of sintered (in vacuum) $NbC_{0.98}$ (contents: non-combined C – 0.08%,

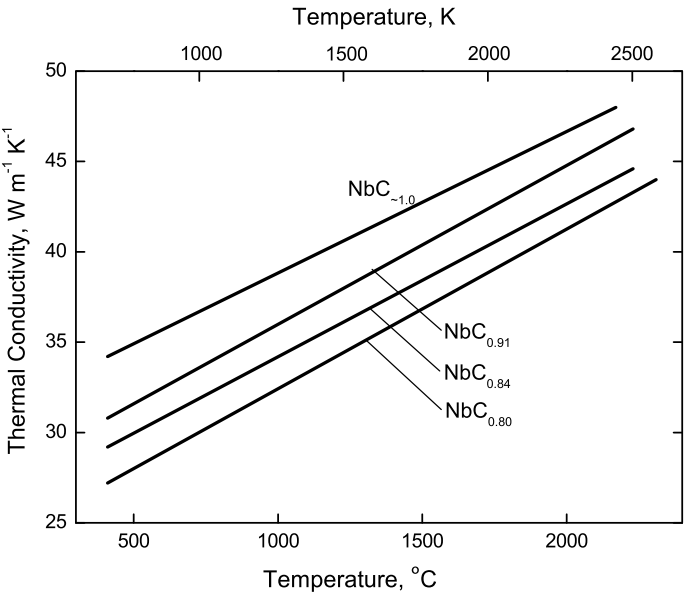

Fig. 4.8 Variations of thermal conductivity with temperature for niobium monocarbide NbC_{1-x} phases having different deviations from the stoichiometry (x = 0, 0.09, 0.16, 0.20, corrected to the poreless state [226, 227])

N – 0.03%, O – 0.02%) in the temperature range of 25–2030 °C the following equation:

$$\alpha_m = 6.35 \times 10^{-6} + (0.52 \times 10^{-9})(t - 25), \tag{4.47}$$

for non-stoichiometric phases with x from 0.076 to 0.298 studied by Kempter and Storms [36] in the temperature range of 25–600 °C, the equations modified to the same form could be presented by the following way:

for $NbC_{0.924}$

$$\alpha_m = 5.22 \times 10^{-6} + (1.76 \times 10^{-9})(t - 25), \tag{4.48}$$

for $NbC_{0.825}$

$$\alpha_m = 3.30 \times 10^{-6} + (4.77 \times 10^{-9})(t - 25), \tag{4.49}$$

for $NbC_{0.766}$

$$\alpha_m = 5.11 \times 10^{-6} + (1.58 \times 10^{-9})(t - 25), \tag{4.50}$$

Table 4.5 Average coefficients of linear thermal expansion α_m of near-stoichiometric niobium monocarbide NbC_{1-x} in various temperature ranges

Temperature range, °C	α_m, 10^{-6} K^{-1}	Reference
20–200	5.93[a]	[1031]
20–300	6.04[a]	[1031]
20–400	6.14[a]	[1031]
20–500	6.25[a]	[1031]
	6.52	[177, 1032]
	6.65[b]	[180, 187]
	6.65 ± 0.28[c]	[86]
20–600	6.23[d]	[36]
	6.35[a]	[1031]
	6.70[b]	[180, 187]
	7.90	[188, 189]
20–700	6.45[a]	[1031]
	6.76[b]	[180, 187]
	6.76 ± 0.23[c]	[86]
20–730	6.52	[880, 881]
20–800	6.56[a]	[1031]
	8.00	[188, 189]
20–850	6.84[b]	[180, 187]
	6.84 ± 0.21[c]	[86]
20–900	5.65[e]	[699]
	6.66[a]	[1031]
	6.75[d]	[59]
20–1000	6.55[f]	[192]
	6.60	[7, 178]
	6.76[a]	[1031]
	6.88[b]	[180, 187]
	7.02	[177]
	7.07	[177, 1032]
	7.10[g]	[209]
	7.11[h]	[186]
	8.10	[188, 189]
	8.55	[80]
	9.31	[2]
200–1000	7.77[i]	[190]
20–1100	6.50	[63, 121]
	6.86[a]	[1031]
20–1200	6.50	[170, 694]
	6.90[b]	[180, 187]
	6.97[a]	[1031]
	7.17	[177]
	8.15	[188, 189]
20–1400	6.65	[65, 722]
	6.92[b]	[180, 187]
	7.18[a]	[1031]
	7.20	[299]
	7.29	[177]
	7.30	[170]
	8.20	[188, 189]

(continued)

Table 4.5 (continued)

Temperature range, °C	α_m, 10^{-6} K^{-1}	Reference
25–1450	7.29[j]	[75, 104]
20–1500	7.28[a]	[1031]
	7.34[f]	[192]
	7.46	[177, 1032]
	8.10[h]	[186]
20–1600	7.15[b]	[180, 187]
	7.38[a]	[1031]
	8.30	[188, 189]
20–1730	7.25	[5]
	7.52	[880, 881]
20–1800	7.41[b]	[180, 187]
	7.58[a]	[1031]
	8.40	[188, 189]
25–1925	6.30[k]	[181]
20–2000	7.38[j]	[184]
	7.79[a]	[1031]
	7.57[b]	[180, 187]
	7.86[f]	[192]
	8.27[h]	[186]
	8.50	[188, 189]
20–2130	7.95	[880, 881]
20–2200	8.01[a]	[1031]
	8.60	[188, 189]
20–2300	8.12[a]	[1031]
	7.76[b]	[180, 187, 727]
20–2400	8.22[a]	[1031]
	8.37[f]	[192]
20–2450	8.40[l]	[193]
20–2500	8.33[a]	[1031]
	8.75[h]	[186]
20–2600	7.92[b]	[180, 187]
	8.43[a]	[1031]
	8.72[f]	[192]
20–2700	8.54[a]	[1031]
20–2850	8.99[h]	[186]
20–2900	8.02[b]	[180, 187]

[a]Calculated on the basis of approximation function
[b]Hot-pressed $NbC_{0.97}$ (13–18% porosity), measured by telemicroscope (dilatation) method in vacuum
[c]$NbC_{0.88}$, measured by high-temperature X-ray diffraction
[d]$NbC_{0.92}$, measured by high-temperature X-ray diffraction
[e]$NbC_{\sim 1.0}$, measured by high-vacuum high-temperature X-ray diffraction
[f]Hot-pressed $NbC_{0.95}$, measured by using an optical micrometer in Ar atmosphere
[g]Hot-pressed $NbC_{0.98}$, measured by a dilatometric method in Ar atmosphere
[h]Hot-pressed $NbC_{\sim 1.0}$ (8% porosity), measured by using an optical micrometer in Ar atmosphere

(continued)

Table 4.5 (continued)

[i]Sintered (in vacuum) $NbC_{0.93}$, measured by high-temperature X-ray diffraction and dilatometric methods
[j]$NbC_{0.98}$, measured by high-temperature X-ray diffraction
[k]Hot-pressed (2% porosity), measured by a dilatometric method
[l]Hot-pressed $NbC_{0.97}$, measured by a dilatometric method

for $NbC_{0.702}$

$$\alpha_m = 4.98 \times 10^{-6} + (1.59 \times 10^{-9})(t - 25), \tag{4.51}$$

where α_m is the average coefficient of linear thermal expansion, K^{-1} and t is temperature, °C. The influence of carbon content on the thermal expansion of monocarbide NbC_{1-x} and semicarbide $Nb_{2+x}C$ phases within their homogeneity ranges can be seen in Figs. 4.3 and 4.9. The average values of coefficient of linear thermal expansion of near-stoichiometric niobium semicarbide Nb_2C were evaluated as $(7.0 \pm 0.3) \times 10^{-6}$ K^{-1} in the range of temperatures from 12 to 190 °C and 6.5×10^{-6} K^{-1} in the range of 20–1100 °C [62]. On the basis of X-ray measurements Lönnberg and Lundström [47] proposed for α_m, K^{-1}, along the main crystallographic directions of semicarbide modifications the following equations:

for α-Nb_2C (in the temperature range of 25–1200 °C)

$$\alpha_{a,m} = 3.093 \times 10^{-6} + (9.761 \times 10^{-10})(t + 273), \tag{4.52}$$

$$\alpha_{b,m} = 9.424 \times 10^{-6} + (1.551 \times 10^{-9})(t + 273), \tag{4.53}$$

$$\alpha_{c,m} = 5.327 \times 10^{-6} + (1.381 \times 10^{-9})(t + 273), \tag{4.54}$$

for β-$Nb_{2+x}C$ (in the temperature range of 25–1300 °C)

$$\alpha_{a,m} = 4.942 \times 10^{-6} + (1.855 \times 10^{-9})(t + 273), \tag{4.55}$$

$$\alpha_{c,m} = 3.884 \times 10^{-6} + (3.012 \times 10^{-9})(t + 273), \tag{4.56}$$

where $\alpha_{i,m}$ is the average coefficient of linear thermal expansion along the i-th direction, K^{-1} and t is temperature, °C. The thermal expansion of α-Nb_2C is strongly anisotropic, while that of β-$Nb_{2+x}C$ is nearly isotropic; the volume thermal expansion of the former phase is, however, slightly larger than that of the latter. Some recent works [152, 356, 402, 409, 738, 955] are devoted to the calculations of coefficients of thermal expansion of niobium carbides by means of theoretical modelling.

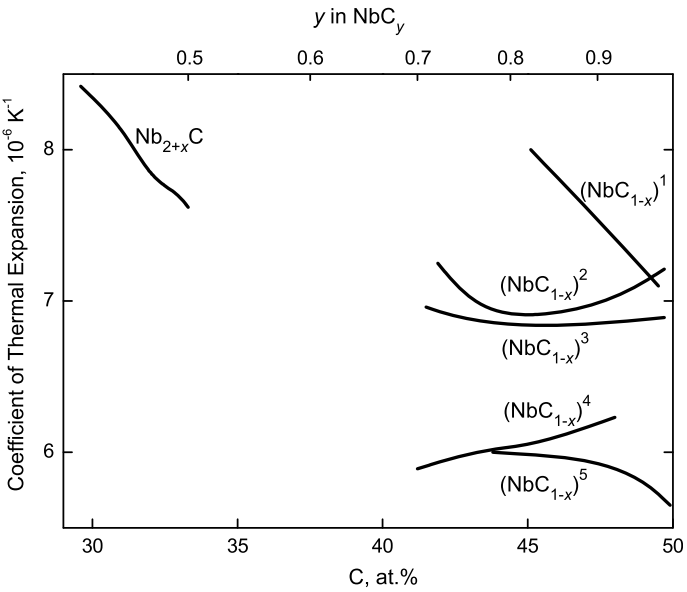

Fig. 4.9 Coefficients of linear thermal expansion in the temperature range of 20–1000 °C (when it is not indicated specially) of niobium monocarbide NbC_{1-x} (1 – hot-pressed, dilatometric method [209]; 2 – [1, 67]; 3 – hot-pressed, 6–10% porosity, content non-combined C $\leq$ 0.16%, dilatometric method [10, 11, 185, 191, 854]; 4 – in the range of 20–600 °C [36]; 5 – in the range of 20–900 °C, high-temperature X-ray diffraction method [699]) and semicarbide $Nb_{2+x}C$ [1, 51, 67] within their homogeneity ranges as functions of carbides composition

In comparison with other ultra-high temperature materials the values of thermal conductivity and thermal expansion of niobium carbides in the wide range of temperatures are summarized in Addendum.

4.3 Electro-magnetic and Optical Properties

At room temperature the value of specific electrical resistance (resistivity) of near-stoichiometric niobium monocarbide NbC_{1-x} lies within the area of 0.2–0.8 μΩ m [1, 5, 62–65, 80, 159, 176–178, 189, 193, 196, 197, 300, 651, 719, 720, 849, 878, 882], and that of near-stoichiometric niobium semicarbide α-Nb_2C – within the area of 0.6–1.4 μΩ m [1, 116]. The variation of this property with temperature for near-stoichiometric monocarbide NbC_{1-x} is shown on the basis of several sources in Fig. 4.10. In the wide temperature range from 200–500 °C up to 2600–2900 °C the resistance of monocarbide phase enlarges with increasing temperature, practically in accordance with linear relationship; that is an evidence of mainly metallic type of conduction in NbC_{1-x} [10, 11]. For the temperature range from 20 to 2400 °C the following equation was recommended by Maltseva et al.

[62, 170, 756] for the resistivity of near-stoichiometric niobium monocarbide NbC_{1-x}, ρ, μΩ m (corrected to the poreless state):

$$\rho = 0.51 + (4.0 \times 10^{-4})(t - 20), \tag{4.57}$$

where t is temperature, °C. Dy and Williams [210] applied to niobium monocarbide phase the empirical equation, which was experimentally determined for ultra-low and low temperature range (4–300 K) by Allison et al. [211] as follows:

for single crystal $NbC_{0.98}$

$$\rho(T) = 24.0 + (2.2 \times 10^{-2})T + 22.9\exp(-278/T), \tag{4.58}$$

and for single crystal $NbC_{0.87}$

$$\rho(T) = 121.6 + (5.0 \times 10^{-3})T + 9.5\exp(-474/T), \tag{4.59}$$

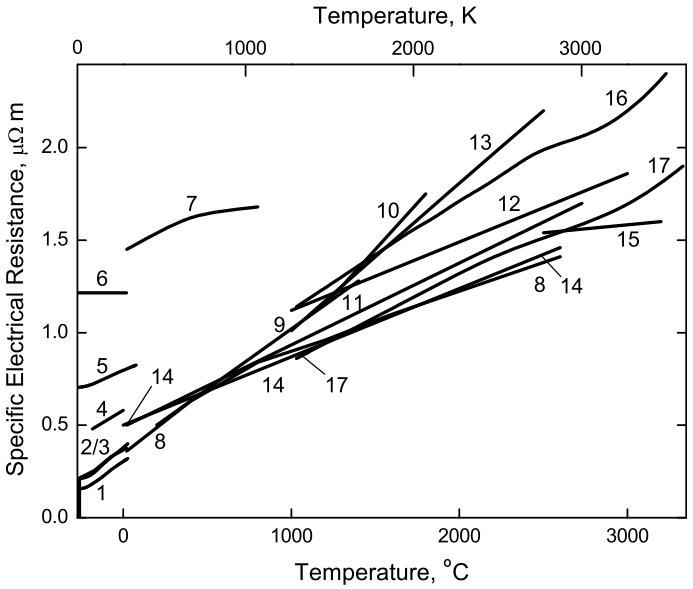

Fig. 4.10 Variation of specific electrical resistance with temperature for niobium monocarbide NbC_{1-x} materials on the basis of several sources: 1 – hot-pressed $NbC_{0.98}$, 3% porosity [183, 201]; 2 – 14% porosity [177, 189, 200, 1033]; 3 – single crystal $NbC_{0.975}$ <100> [971]; 4 – sintered $NbC_{0.95}$, 23% porosity [176]; 5 – single crystal $NbC_{0.95}$ [758]; 6 – single crystal $NbC_{0.84}$ <100> [971]; 7 – single crystal $NbC_{0.82}$ [179]; 8 – [10, 11, 199]; 9 – hot-pressed and annealed, almost poreless $NbC_{0.99}$ [198]; 10 – hot-pressed [67]; 11 – based on several sources [5]; 12 – hot-pressed $NbC_{0.97}$, 13–18% porosity, corrected to the poreless state [180]; 13 – sintered and annealed in vacuum, 2% porosity [27, 759]; 14 – hot-pressed and annealed $NbC_{0.87}$ [170, 756]; 15 – [120]; 16 – sintered $NbC_{0.91–0.97}$, 15–17% porosity, contents: non-combined C – 0.35–0.63%, N $\leq$ 0.26%, O – 0.07–0.21% [243]; 17 – sintered $NbC_{0.91–0.97}$ (*see also* 16), corrected for porosity [765] (when it is not indicated specially, data are given for near-stoichiometric compositions)

Table 4.6 Average thermal coefficients of resistivity α_R of niobium monocarbide NbC_{1-x} in various temperature ranges[a]

Temperature range, °C	α_R, 10^{-3} K^{-1}	Reference
(−183)–0	1.14[b]	[176]
20–1000	~1.35	[67, 206]
	~0.80[c]	[67, 206]
	0.52[d]	[757]
	~0.20[e]	[67, 206]
100–1100	1.75	[1, 198, 693]
300–2300	0.86[f]	[5, 10, 207, 693]

[a]When it is not indicated specially, value reported is for near-stoichiometric composition
[b]Sintered materials, 23% porosity, carbon content – 48.7 at.%
[c]Hot-pressed materials, carbon content – 47.4 at.%, corrected for porosity
[d]Hot-pressed materials, carbon content – 48.9 at.%, thermal coefficient $\alpha = 1/\rho_\theta \times (d\rho/dT)_\theta$, corresponding to Debye temperature $\theta = 805$ K ($(d\rho/dT)_\theta = 3.5 \times 10^{-4}$ μΩ m K^{-1}), corrected for porosity
[e]Hot-pressed materials, carbon content – 42.9 at.%, corrected for porosity
[f]Hot-pressed materials, $d\rho/dT = 9.0 \times 10^{-4}$ μΩ m K^{-1}

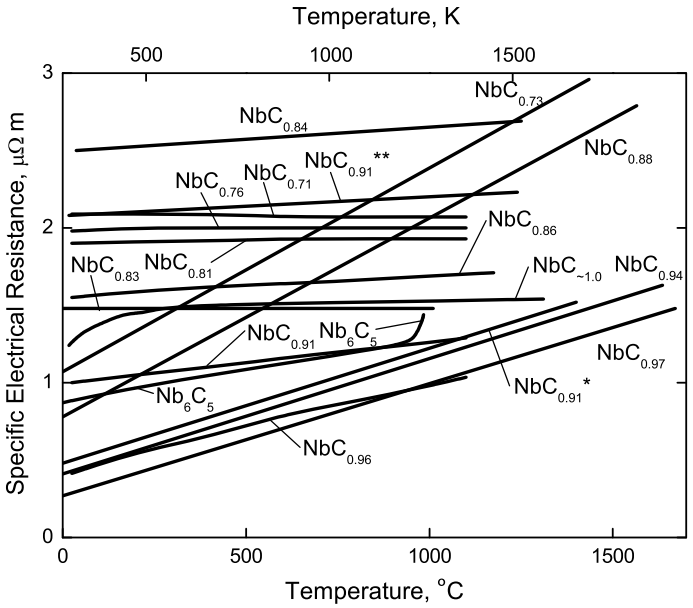

Fig. 4.11 Variations of specific electrical resistance with temperature for the NbC_{1-x} phases having different deviations from the stoichiometry ($x = 0.04$, 0.09, 0.14, 0.19, 0.24, 0.29 [202]; annealed 100 h at 2400 °C, $x = 0.03$, 0.06, 0.09*, 0.12, 0.27 [5]; $x = 0$, 0.09**, 0.16, corrected to the poreless state [226]; disordered $x = 0.17$ and ordered Nb_6C_5 [74, 210])

where T is temperature, K (the resistivity of single crystal $NbC_{0.76}$ is nearly constant over 4–300 K temperature range varying only from 1.53 to 1.55 μΩ m [211]). In the interval from $T = 0$ to $T = 350$ K the single crystal $NbC_{0.95}$ resistivity measured by Allison et al. [758] can be described by the Wilson expression (in μΩ m):

$$\rho(T) = 0.705 + (1.606 \times 10^{-8})T^3 \int_0^{216/T} x^3 / \sinh^2 x dx \qquad (4.60)$$

which can be simplified for $T \leq$ 40–70 K up to $\rho(T) = 0.705 + AT^3$, where A is a constant. The data on the thermal coefficients of resistivity α_R of near-stoichiometric NbC_{1-x} in the various temperature ranges are presented in Table 4.6. The variations of resistivity with temperature for the phases having different deviations from the stoichiometry, including the ordered phase Nb_6C_5 as an especial example, are presented in Fig. 4.11 on the basis of several sources. The specific electrical resistance of NbC_{1-x} at the melting point is 2.54 μΩ m [62, 159, 177]. Within the homogeneity range the resistivity of $Nb_{2+x}C$ and NbC_{1-x} supplements with increasing carbon deficit in non-metal sublattice, especially the considerable growth of resistivity is inherent for the monocarbide phase, that is shown in Fig. 4.12, where the data from several sources are collected. The variation

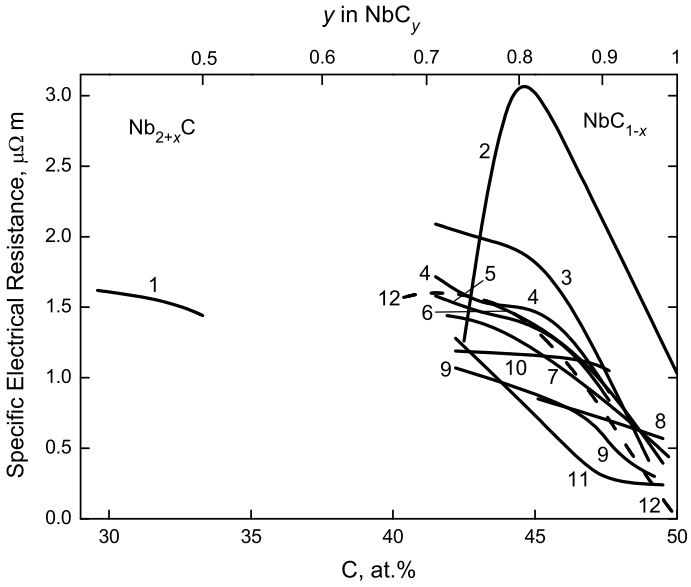

Fig. 4.12 Variations of specific electrical resistance at room temperature within the homogeneity ranges of niobium semicarbide $Nb_{2+x}C$ (1 – sintered [51]) and monocarbide NbC_{1-x} (on the basis of several sources: 2 – [227]; 3 – [204, 215]; 4 – sintered in vacuum, 5–15% porosity [171, 845], 5 – [205]; 6 – single crystals [211]; 7 – [53, 202, 208]; 8 – hot-pressed [209]; 9 – [5]; 10 – hot-pressed and annealed, content non-combined C – 0.15–0.52% [218, 848]; 11 – [203]; *in dash*: 12 – residual resistivity at ultra-low temperatures, single crystals materials [971])

of $d\rho/dT$ with composition at ambient temperatures is shown in Fig. 4.13; the increase in carbon vacancy concentration in the NbC_{1-x} phase leads to a change not only in absolute resistivity, but also – a regular decrease in $d\rho/dT$ and α_R values [10, 11, 80, 202]. The superconducting transition temperature T_c of near-stoichiometric monocarbide NbC_{1-x} is 6–12 K [84, 92, 183, 212–214, 230–238, 334, 847, 852, 853, 879, 895, 898, 906, 1020, 1036]; it varies with composition considerably, e.g. falling from 11.1 K for $NbC_{0.98}$ to 1.05 K for $NbC_{0.83}$. The highest temperature is reached for the most stoichiometric composition, so any deviation therefrom results in a lowering and eventual disappearance (<1.05 K at x = 0.2–0.3) of the superconductivity in general [212–214]. The transition temperatures T_c of semicarbide phases: for $Nb_{2.08}C$ <1.6 K [213], for $Nb_{2.0}C$ ~9.2 K [212, 933], for $Nb_{1.96}C$ <2 K [212, 1035]. At room temperature the Hall and Seebeck coefficients of near-stoichiometric NbC_{1-x} are $R = -(0.8–1.3) \times 10^{-10}$ m^3 A^{-1} s^{-1} and $S = -(3–6)$ μV K^{-1}, respectively [1, 7, 10, 11, 67, 82, 216, 217, 760, 849]; the concentration variations of values of these constants within the homogeneity range of niobium monocarbide NbC_{1-x} are presented in Figs. 4.14 and 4.15.

Near-stoichiometric niobium monocarbide NbC_{1-x} is a paramagnetic substance with molar magnetic susceptibility χ_m (SI) ≈ (100–260) × 10^{-6} cm^3 mol^{-1} at room temperature very slightly varying with temperature in the range from −200 to 900 °C [1, 10, 11, 67, 74, 218–225, 850, 851]. The weak temperature influence and small absolute value of susceptibility indicate that the magnetic properties of

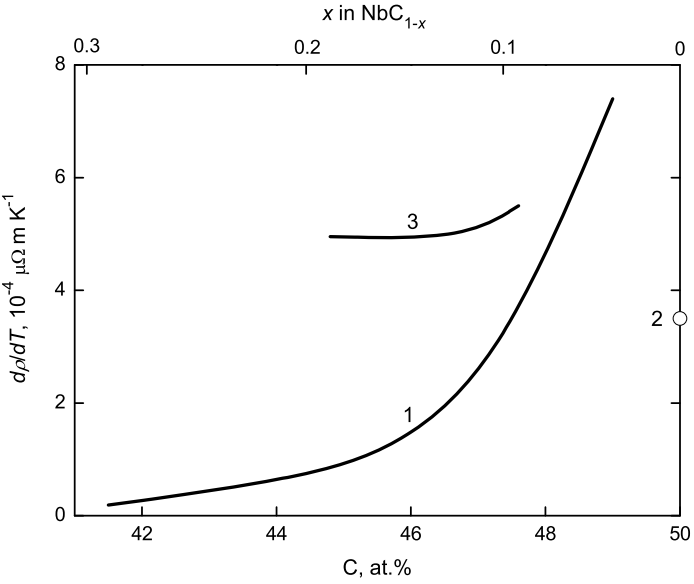

Fig. 4.13 Variation of $d\rho/dT$ (slope of resistivity vs. temperature) at ambient temperatures within the homogeneity range of niobium monocarbide NbC_{1-x} (1 – [10, 11, 202]; 2 – [80]; 3 – hot-pressed, content non-combined C – 0.23–0.52% [218, 848])

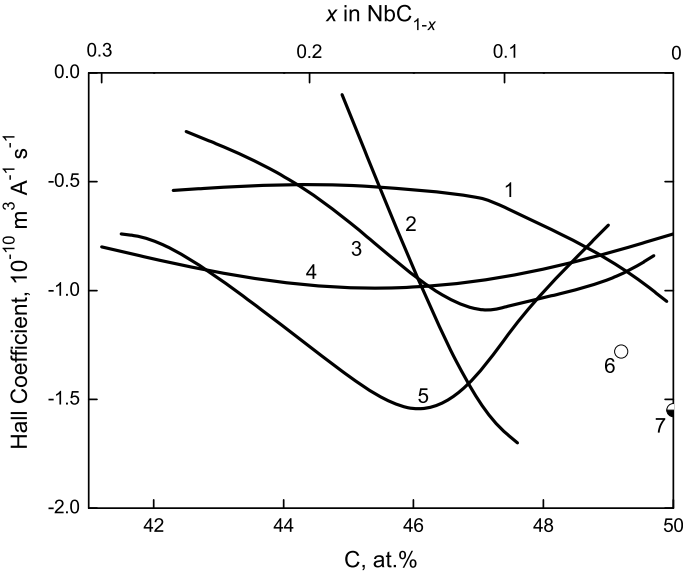

Fig. 4.14 Variations of Hall coefficient at room temperature within the homogeneity range of niobium monocarbide NbC_{1-x} on the basis of several sources (1 – sintered [208, 227]; 2 – hot-pressed, content non-combined C – 0.23–0.52% [218, 848]; 3 – [10, 11]; 4 – single crystal materials [971]; 5 – sintered (in vacuum) [215]; 6 – [760]; 7 – hot-pressed [65, 761, 934, 935])

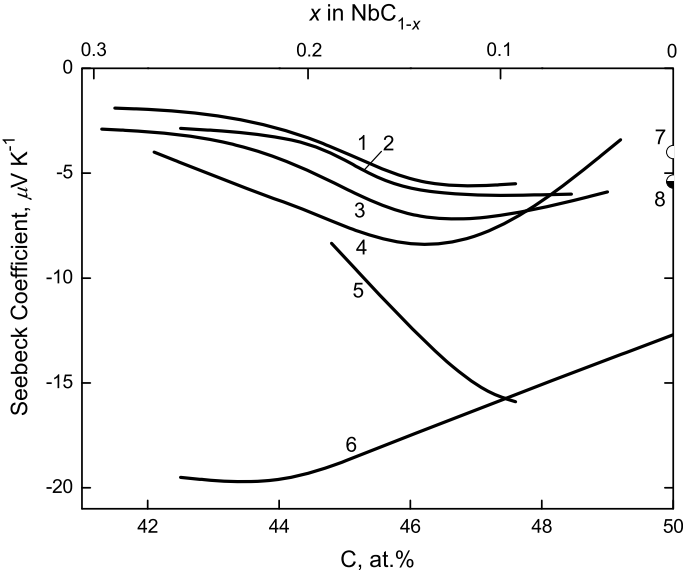

Fig. 4.15 Variations of Seebeck coefficient at room temperature within the homogeneity range of niobium monocarbide NbC_{1-x} on the basis of several sources (1 – sintered in vacuum, 5–15% porosity [171]; 2 – sintered in vacuum, 2–3% porosity [278]; 3 – [215]; 4 – [203]; 5 – hot-pressed, content non-combined C – 0.23–0.52% [218, 848]; 6 – [227]; 7 – hot-pressed [216]; 8 – [761])

monocarbide NbC_{1-x} are determined primarily by conducting electrons [10, 11]. With increasing carbon deficit within the homogeneity range the value of magnetic susceptibility of monocarbide declines and reaches its diamagnetic (negative) minimum at about $-(50–250) \times 10^{-6}$ cm^3 mol^{-1}, corresponding to $\sim NbC_{0.8}$ composition (*see* Fig. 4.16), and then it enlarges again proceeding to the paramagnetic region [10, 11, 225]. Gusev and Rempel [74, 222] showed that the minima of χ in niobium monocarbide NbC_{1-x} phases in disordered and ordered states occur for the composition of $NbC_{0.83}$ at $-(105–125) \times 10^{-6}$ cm^3 mol^{-1}, and for $NbC_{0.81}$ at $-(200–220) \times 10^{-6}$ cm^3 mol^{-1}, respectively. The temperature coefficient of magnetic susceptibility $d\chi/dT$ changes its sign with composition; it is positive for compositions around $NbC_{0.78}$ and negative – for near-stoichiometric composition $NbC_{0.98}$ [10, 11, 228]. According to Caudron et al. [229], niobium semicarbide $Nb_{2\pm x}C$ is also a paramagnetic substance with the value of its mass magnetic susceptibility – $\sim 485 \times 10^{-6}$ cm^3 mol^{-1}.

The reflectance spectra from the ultraviolet wavelength region to the mid-infrared band of single crystal monocarbide NbC_{1-x} materials with various carbon content are shown in Fig. 4.17; specular reflectance was measured between 0.025 and 11 eV, and ellipsometry measurements were made at 1.96 eV as well to conclude that HIPed materials had optical and electric properties that were almost identical to those of single crystals with nearly the same composition [211]. Optical constants (refractive index n and extinction coefficient k versus photon energy

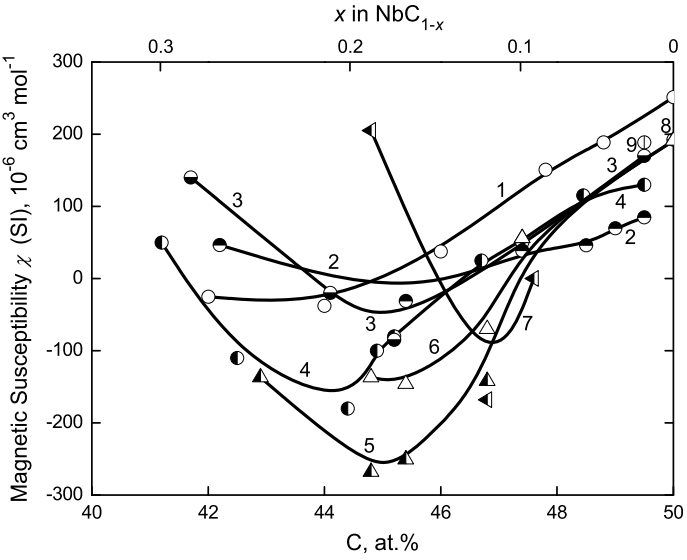

Fig. 4.16 Variation of molar magnetic susceptibility at room temperature within the homogeneity range of niobium monocarbide NbC_{1-x} on the basis of several sources: 1 – from 48.8 to 50 at.% C the values extrapolated [219, 763]; 2 – [220]; 3 – [762]; 4 – [221]; 5 – ordered and 6 – disordered carbide phases [222, 224]; 7 – content non-combined C – 0.23–0.52% [218, 848]; 8 – [761]; 9 – [764]

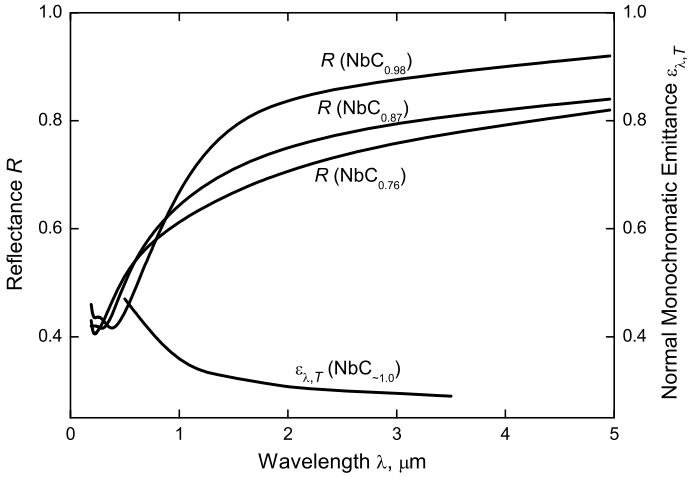

Fig. 4.17 Variations of reflectance R ($NbC_{0.76}$, $NbC_{0.87}$ and $NbC_{0.98}$ single crystal materials [211]) and normal monochromatic emittance $\varepsilon_{\lambda,T}$ ($NbC_{\sim 1.0}$ pyrolytic coating on graphite substrate at 1530–2220 °C [10, 11, 247, 1042]) with wavelength λ

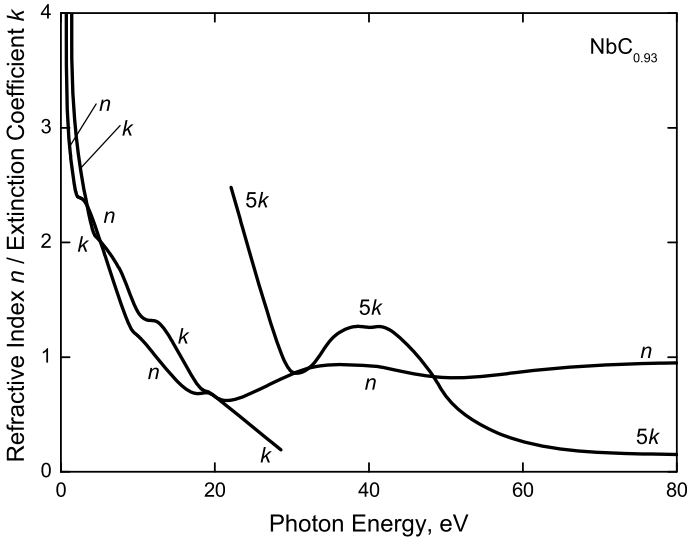

Fig. 4.18 Refractive index n and extinction coefficient k of single crystal $NbC_{0.93}$ deduced through a Kramers-Kronig analysis of the reflectance spectrum (k is shown in an expanded scale $\times 5$ above 22 eV) [239, 972]

relationships) deduced through a Kramers-Kronig analysis of the reflectance spectrum of single crystal $NbC_{0.93}$ are presented in Fig. 4.18. Optical spectra of NbC_{1-x} in the infrared (IR) and visible ranges as well as X-ray emission and absorption spectra were considered by Alyamovskii et al. [241], Ramqvist et al. [240], Zhurakovskii et al. [1039–1041], Korsunskii et al. [1038], Kammori et al. [1043] and Upadhyaya [11]. The absorption wave numbers of basic maxima of IR band for niobium carbides are 1100 cm^{-1} (for $NbC_{0.98}$), 1075 cm^{-1} (for $NbC_{0.77}$) and 1080 cm^{-1} (for Nb_2C) [241]. At the common conditions the colour of niobium monocarbide NbC_{1-x} powder is light-brown, gray-brown or sometimes gray with a lavender/violet tint, golden-orange colour is more inherent to near-stoichiometric compositions NbC_{1-x} in the compact state [1, 6, 7, 10, 11, 63, 67, 84, 117, 882].

The normal monochromatic emittance (spectral emissivity) $\varepsilon_{\lambda,T}$ of niobium monocarbide NbC_{1-x} falls noticeably with wavelength λ growth (*see* Fig. 4.17) and only slightly varies with temperature [1, 10, 11, 247, 715]; the linear relationship for $\lambda = 0.65$ μm

Table 4.7 Normal monochromatic emittance ε_λ (λ = 0.630–0.665 μm) of niobium monocarbide NbC_{1-x} materials in various temperature ranges

Monochromatic emittance, ε_λ	Temperature range, °C	Reference
0.28	725	[200]
0.35–0.41[a]	900–1850	[242, 715]
0.42–0.49[b]	900–3000	[247]
0.55–0.62[c]	2370–2990	[62, 120]
0.59–0.66[d]	1000–3200	[243, 246, 765]
0.62–0.66[b]	1000–2200	[1]
0.71–0.79[e]	1430–1930	[248]
0.73[f]	2000–2500	[10]
0.75[g]	1500–2400	[250]
0.85[h]	800–1800	[5, 63, 196, 244, 245]

[a]Arc-melted (via floating zone) polycrystalline $NbC_{0.83\pm0.01}$, measured in vacuum 10^{-5}–10^{-6} Pa; ε_λ falls with the temperature growth linearly (slope is $\sim 6.6 \times 10^{-5}$ K^{-1})

[b]Sintered materials; ε_λ falls with the temperature growth linearly

[c]ε_λ falls with the temperature growth

[d]Sintered materials $NbC_{0.91–0.97}$ (15–17% porosity, mean grain size ≤15–16 μm, contents: non-combined C – 0.35–0.63%, N ≤ 0.26%, O – 0.07–0.21%); ε_λ falls with the temperature growth linearly

[e]For λ = 0.81 μm, sintered materials with 32% porosity, the surface roughness effect on ε_λ is very low, after high-temperature exposure in vacuum ε_λ gradually decreases from 0.79 and approaches a constant value of 0.71, as a result of exposure to hot H_2 at 1260 °C for 2 h ε_λ reduces from 0.79 to 0.72 [249]

[f]Hot-pressed materials

[g]Materials with composition $NbC_{0.77}$, deposited from gas phase and annealed in vacuum at 2200 °C for 1 h exposure

[h]Powdered materials on tantalum substrate

$$\varepsilon_\lambda = 0.4913 - (6.6 \times 10^{-5})(t + 273), \tag{4.61}$$

where t is temperature, °C, was obtained by Mackie et al. [242, 715] for arc-melted (via floating zone) polycrystalline monocarbide $NbC_{0.83\pm0.01}$ in the temperature range 900–1850 °C. For near-stoichiometric NbC_{1-x} Petrov [10, 246] has recommended in the temperature range 1100–3100 °C the equation as follows:

$$\varepsilon_\lambda = 0.66 - (2.60 \times 10^{-5})(t - 727), \tag{4.62}$$

where t is temperature, °C. According to the experimental work by Okhremchuk et al. [766], emissivity ε_λ of NbC_{1-x} phases decreases from 0.75 to 0.70 with increasing value of index x in NbC_{1-x} formula from 0.02 to 0.18. The data on normal monochromatic emittance ε_λ (λ = 0.630–0.665 μm) and integral (total) emittance ε_T for NbC_{1-x} materials produced by different manufacturing methods and measured at various temperatures are listed in Tables 4.7 and 4.8.

The thermionic emission characteristics (electron work function and Richardson constants) of various niobium carbide phases are given in Table 4.9 and Fig. 4.19. The calculated values of emission current density for near-stoichiometric monocarbide NbC_{1-x} are about 7×10^{-4}, 9×10^{-2} and 0.2 A m^{-2} at 730, 1230 and 1730 °C, respectively [5]. Using the photoemission technique, Lindberg and Johansson [875] revealed that the work function of $NbC_{0.83}$ (100) increases dramatically with increasing the exposure to O_2, while only small increases in the work function are observed after the exposure to CO media. A typical brightness characteristic (U = 75 kV) of niobium monocarbide cathode (thermionic emitter for electron microscopy) is shown in Fig. 4.20. Field emission properties of niobium monocarbide NbC_{1-x} materials were reported in several works [255–257, 776, 811, 1045–1047].

Table 4.8 Integral (total) emittance ε_T of near-stoichiometric niobium monocarbide NbC_{1-x} in various temperature ranges

Integral emittance, ε_T	Temperature range, °C	Reference
0.40–0.55[a]	1000–3200	[243, 246, 765]
0.42–0.46[b]	730–2430	[247]
0.57–0.58[c]	2000–2500	[1, 10, 11]

[a]Hemispherical emittance ε_T for sintered $NbC_{0.91-0.97}$ (15–17% porosity, mean grain size ≤15–16 μm, contents: non-combined C – 0.35–0.63%, N ≤ 0.26%, O – 0.07–0.21%); ε_T increases with the temperature growth (from 1000 to 3000 °C the relationship ε_T – T is linear)
[b]Normal emittance ε_T for sintered materials; ε_T increases with the temperature growth
[c]Normal emittance ε_T for hot-pressed materials; ε_T increases with the temperature growth

Table 4.9 Thermionic emission characteristics (electron work function and Richardson constant) of niobium carbide phases

Composition	Work function[a], $\varphi = \varphi_0 + (d\varphi/dT)_{av}T$, eV	Richardson constant, A, 10^4 A m^{-2} K^{-2}	Temperature range, °C	Remarks[b]	Reference
Nb_2C	4.10[c]	–	1500	TiM, hot-pressed	[271]
	4.12[c]	–	1500	TiM, hot-pressed	[10, 11, 1044]
	5.20	–	50	–	[824]
	$5.50 - (7.7 \times 10^{-4})T$	–	1200–1800	TiM, hot-pressed	[1, 191, 1044]
$NbC_{0.72}$	$4.32 - (1.8 \times 10^{-4})T$	–	1200–1750	TiM, hot-pressed	[1, 191]
$NbC_{0.75}$	4.00[c]	–	1500	TiM, hot-pressed	[10, 11, 1044]
$NbC_{0.81}$	$4.45 - (1.7 \times 10^{-4})T$	–	1200–1800	TiM, hot-pressed	[1, 191]
$NbC_{0.82}$	4.15[c]	–	1500	TiM, hot-pressed	[10, 11, 1044]
	4.40	–	50	–	[824]
	$4.44 - (1.6 \times 10^{-4})T$	–	1200–1800	TiM, hot-pressed	[1, 191]
$NbC_{0.86}$	3.84 ($d\varphi/dT = 0$)	–	1000–1700	TiM, single crystal (100), measured in vacuum ($\sim 1.3 \times 10^{-6}$ Pa)	[268, 857, 858]
	3.90 ($d\varphi/dT = 0$)	–	1000–1700	TiM, single crystal (102), measured in vacuum ($\sim 1.3 \times 10^{-6}$ Pa)	[268, 857, 858]
	$3.95 - (6.48 \times 10^{-5})T$	–	≤1700	TiM, single crystal (100)	[254, 767]
	$3.95 - (2.93 \times 10^{-5})T$	–	≤1700	TiM, single crystal (102)	[254, 767]
	4.0	–	1000–1700	CPDM, single crystal (100), measured in vacuum ($\sim 1.3 \times 10^{-6}$ Pa)	[254, 268, 767]
	4.1	–	25	CPDM, single crystal (102)	[254, 767]
$NbC_{0.88}$	4.05[c]	–	1500	TiM, hot-pressed	[10, 11, 1044]
	$4.37 - (1.8 \times 10^{-4})T$	–	1150–1800	TiM, hot-pressed	[1, 191]
$NbC_{0.90}$	4.05[c]	–	1400	TiM, sintered	[1, 254]
$NbC_{0.91}$	$4.30 - (1.7 \times 10^{-4})T$	–	1150–1750	TiM, hot-pressed	[1, 191]
$NbC_{0.92}$	4.00[c]	–	1500	TiM, hot-pressed	[10, 11, 1044]
$NbC_{0.95}$	4.2		1600	UPS, single crystal (100)	[257]
	5.2		1600	UPS, single crystal (111)	[257]
$NbC_{0.99}$	3.93[c]	–	1500	TiM, hot-pressed	[10, 11, 1044]
	4.21 ± 0.05	–	25	CPDM (W), hot-pressed	[1, 261]
	$4.25 - (1.9 \times 10^{-4})T$	–	1150–1750	TiM, hot-pressed	[1, 191]
$NbC_{\sim 1.0}$	~2.24	$\sim 10^{-6}$	–	TiM	[1, 254]
	3.38[d]	–	–	Single crystal (001), theoretically evaluated	[395]
	3.50	–	–	TiM, pyrolytic coating	[1, 265]
	3.58[c]	–	1230	TiM, powder	[1, 258]
	3.66[e]	–	–	Single crystal (001), theoretically evaluated	[395]

(continued)

Table 4.9 (continued)

Composition	Work function[a], $\varphi = \varphi_0 + (d\varphi/dT)_{av}T$, eV	Richardson constant, A, 10^4 A m^{-2} K^{-2}	Temperature range, °C	Remarks[b]	Reference
	3.70 ± 0.05[c]	–	–	TiM, metal-carburized	[1, 254]
	3.70 ± 0.1	–	–	TiM, metal-carburized	[1, 254]
	3.74[c]		1130	TiM, powder	[259]
	3.81[c]	418	2130	TiM, metal-carburized wire	[253]
	3.83[c]	–	2100	TiM, hot-pressed	[263, 264]
	3.85	–	–	Calculated value (unrelaxed surface)	[949, 950]
	3.95[f]	–	–	Single crystal (100), theoretically evaluated	[894]
	4.02[c]	–	25	TiM, powder	[259]
	4.07 ± 0.04[c]	–	1330	TiM, hot-pressed, plain samples	[1, 262]
	4.10	–	–	UPS, single crystal (100)	[875, 949, 950]
	4.10[c]	–	2180	TiM, powder on W	[1, 260]
	4.10 − (2.5 × 10^{-4})T[c]	–	1180–1930	TiM, powder	[1, 10, 11, 254]
	4.14 ± 0.05	–	25	CPDM, hot-pressed	[1, 262]
	4.2[g]	–	25	FEM, single crystal (001), clean surface	[716, 776, 777, 949, 950]
	4.20[c]	72.3	2130	TiM, powder on Ta substrate	[253]
	4.26	–	–	Calculated value (relaxed surface)	[949, 950]
	4.45[h]	–	–	Single crystal (100), theoretically evaluated	[894, 949, 950]
	5.0[i]	–	25	FEM, single crystal (111), clean surface	[716, 776, 777]
	–	–	–	AeE, cold cathode, ∼1 μm fibre, 35 A m^{-2}, 6.7 × 10^{-5} Pa for >10^3 h	[1, 255, 256]

[a]T is temperature, K
[b]Methods applied for the experimental determination of the work function (TiM – thermionic method, UPS – ultraviolet photoemission spectroscopy technique, CPDM – contact potential difference method, second electrode is given in brackets, AeE – autoelectronic emission method) and manufacturing methods for the fabrication of a particular material (or its constitution) are marked
[c]The values of effective electron work function
[d]Calculated using the Perdew's generalized gradient approximation (GGA) for the fixed surface model
[e]Calculated using the Perdew's GGA for the relaxed surface model
[f]Calculated using the full-potential linear-muffin-tin-orbital (FP-LMTO) method
[g]For graphene-covered surface φ = 4.2 eV
[h]Calculated using the linear-muffin-tin-orbital with atomic sphere approximation (LMTO-ASA) technique
[i]For graphene-covered surface φ = 3.8 eV

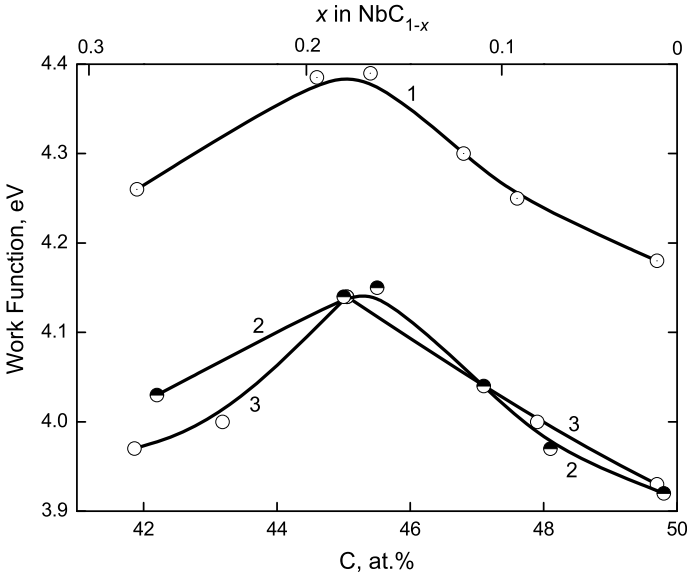

Fig. 4.19 Variations of effective work function of hot-pressed niobium monocarbide NbC_{1-x} materials with carbon content measured by "total current" method: at 50 °C (1 – [824]) and 1530 °C (2 – [854], 3 – [10, 11, 191, 251, 271])

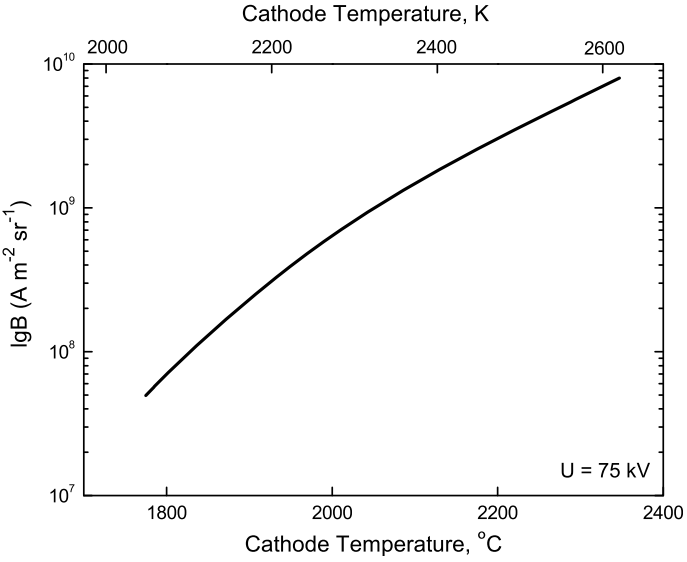

Fig. 4.20 Typical brightness (B) characteristic (U = 75 kV) of niobium monocarbide cathode (thermionic emitter with electron work functions $\varphi = 4.07$ eV, $\varphi_{eff} = 3.81$ eV, Richardson constant $A = 418 \times 10^4$ A m^{-2} K^{-2}, $T = 2400$ K) produced by the carburization reaction of metal wire/foil with benzene vapour [252, 253]

The recommended values of electrical resistivity, magnetic susceptibility, integral and spectral emittances and thermionic emission characteristics (electron work function and Richardson constant) for niobium carbides are given in the wide range of temperatures in comparison with other ultra-high temperature materials in Addendum.

4.4 Physico-mechanical Properties

The physico-mechanical properties of niobium carbides are very sensitive to the deviations from the stoichiometry, but also to the crystallographic directions in the materials; it has been just marked in Sect. 4.1 concerning the slip systems in NbC_{1-x}. At room temperature the hardness *HV*, GPa of near-stoichiometric niobium monocarbide NbC_{1-x} is evaluated as 14.3 ± 1.0 (for $NbC_{0.97}$) [2, 281], 15.0 (for $NbC_{0.99}$, contents: non-combined C – 0.07%, O – 0.05%, N – 0.45%, porosity – 7%, mean grain size – 12 μm, 1 N load) [1, 2, 361], 15.1 (for $NbC_{0.96}$, 200 N load) [53], 16.0 [3, 266, 267], 16.7 ± 1.0 (for $NbC_{0.96}$, content O ≤ 0.05%, 1 N load) [2, 282], 17.0 (200 N load) [1, 53], 17.5 ± 0.8 [1], 18.6 ± 1.0 (for $NbC_{0.97}$, content O + N – 0.1%, 1 N load) [2, 279], 18.9 (for $NbC_{0.96}$, 200 N load) [53], 19.1 (for NbC_{1-x} monocarbide phase in contact with carbon) [773], 19.2 [65, 170], 19.6 [84, 774], 20.0 [770], 20.2 [64], 20.3 ± 0.5 (1 N load) [775], 20.9–21.6 (for nanostructured $NbC_{\sim 1.0}$, porosity – 2%, mean grain size – 76 nm, 100 N load) [321, 902], (21.0–21.6) ± (1.2–2.2) [10, 11, 287], 21.1 [882], 21.3 ± 1.3 (for $NbC_{0.99}$, contents: non-combined C – 0.25%, O – 0.03%, N – 0.01%, 0.5 N load) [1, 67, 772], 21.7 ± 1.3 (0.5 N load) [775], 22.0 (contents: O ≤ 0.01%, N ≤ 0.01%, porosity – 2%, mean grain size – 10 μm) [27], 23.0 ± 1.5 (contents: O ≤ 0.1%, N ≤ 0.1%, porosity – 2%, mean grain size – 10 μm) [27], 23.5 (for $NbC_{0.99}$, 0.5 N load) [769], 23.5 (0.5–1 N load) [5, 177, 189, 286, 716, 771], 24.0 ± 0.95 (0.5 N load) [5, 63] and 24.2 (for $NbC_{\sim 1.0}$ probably contaminated by W) [333]. The hardness of NbC_{1-x} ($x \approx 0$) in Mohs scale is >9 [67, 159], 9–10 [104, 117, 177, 1052] and in Rockwell scale *HRA* is 83 kgf mm^{-2} (0.81 GPa) [2, 67, 289], ~90 kgf mm^{-2} (~0.88 GPa) [338], 91 kgf mm^{-2} (0.89 GPa) [177, 189]. The variations of hardness *HV*/*HK* and microhardness $H\mu$ at room temperature with deviation from the stoichiometry and indenter orientation (anisotropy) within the homogeneity ranges of niobium semicarbide $Nb_{2+x}C$ and monocarbide NbC_{1-x} are demonstrated in Fig. 4.21. The variations of hardness *HV* and microhardness $H\mu$ of near-stoichiometric NbC_{1-x} with temperature based on several resources and presented in Fig. 4.22 indicate mainly the linear characteristic hot-hardness behaviour of monocarbide materials on semi-logarithmic plots *HV*/*HK*/$H\mu$ – 1/*T* (or *T*) [283, 284, 292, 294, 295]. In general, in the temperature range from 0 to 1400 °C the hardness-temperature relationship for NbC_{1-x} materials is described perfectly by the exponential equation [5, 782–784]:

$$HV = 23.5 \exp[-(15.3 \times 10^{-4})t], \tag{4.63}$$

which can be approximated for the lower temperatures (<850 °C) by such linear equation [80] as

$$HV = 29.0 - (2.06 \times 10^{-3})t, \tag{4.64}$$

where HV is Vickers hardness, GPa and t is temperature, °C. Fig. 4.23 demonstrates the microhardness $H\mu$ dependence of single crystal materials on

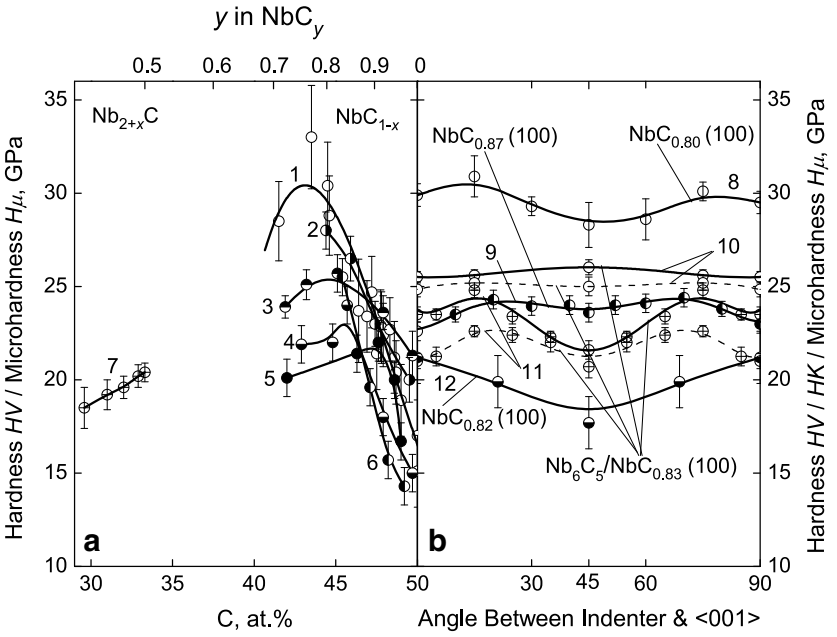

Fig. 4.21 Variations of the hardness HV (1, 8), HK (9–12) and microhardness $H\mu$ (2–7) at room temperature: (**a**) with deviation from the stoichiometry within the homogeneity ranges of niobium monocarbide NbC_{1-x} (1 – produced by gas carbidization and hot-pressing methods, 200 N load [53, 266, 267]; 2 – melted in Ar and annealed in vacuum at 1800–2000 °C, 1.0 N load [277]; 3 – hot-pressed and annealed in vacuum, 6–10% porosity, 0.5 N load [191, 854]; 4 – sintered in vacuum and annealed, porosity – 7–12%, mean grain size – 12–29 μm, contents: non-combined C – 0.05–0.07%, O – 0.05–0.08%, N – 0.27–0.45%, 1.0 N load [1]; 5 – sintered at 2500–2700 °C, 2.0–12.5% porosity, contents: non-combined C ≤ 0.01%, O ≤ 0.05%, 1.0 N load [282]; 6 – produced by carbon diffusion saturation [281]) and semicarbide $Nb_{2+x}C$ phases (7 – sintered in vacuum, 0.5 N load [51]) and (**b**) with indenter orientation (angle between indenter axis and <001> direction) on the (100) surface of single crystal materials (8 – $NbC_{0.80}$, produced by radio-frequency floating zone process, 2 N load [283, 284]; 9 – $NbC_{0.87}$, produced by the floating zone technique (zone melted), 5 N load [297, 1048]; 10 – $Nb_6C_5/NbC_{0.83}$, in long-range ordered state (the two curves for different samples), 1.5 N load [296]; 11 – $Nb_6C_5/NbC_{0.83}$, in short-range ordered state (the two curves for different samples), 1.5 N load [296]; 12 – $NbC_{0.82}$, produced by plasma-arc melting, 1 N load [87])

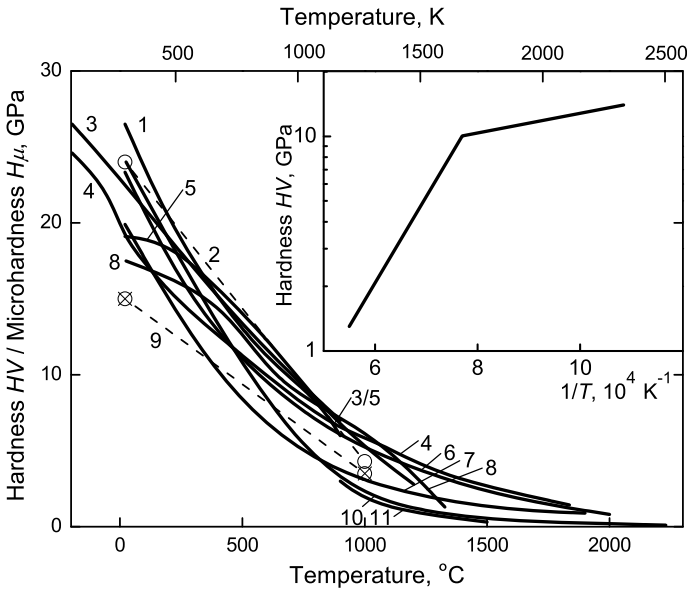

Fig. 4.22 Variations of hardness *HV* (1, 5–6, 10) and microhardness *Hμ* (2–4, 7–9) with temperature for near-stoichiometric niobium monocarbide NbC_{1-x} phases based on several sources: 1 – single crystal $NbC_{0.84}$ [268]; 2 – sintered in vacuum and annealed $NbC_{0.84}$, porosity – 10.5%, mean grain size – 18 μm, 1 N load [1]; 3 – single crystal $NbC_{\sim 1.0}$ (001) produced by plasma-arc melting, 1 N load [273]; 4 – hot-pressed and annealed $NbC_{\sim 1.0}$, porosity – 2–4%, 1 N load (<20 °C) and 0.7 N load (>20 °C) [272, 337, 343]; 5 – sintered with temporary metallic binder (in high vacuum), 3 N load [1053]; 6 – sintered $NbC_{0.95}$, porosity – ∼10% [1, 270]; 7 – hot-pressed $NbC_{\sim 1.0}$ [1]; 8 – single crystal $NbC_{0.80}$ produced by plasma-arc melting, 9.8 N load [285]; 9 – sintered in vacuum and annealed $NbC_{0.99}$, porosity – 7%, mean grain size – 12 μm, 1.0 N load [1]; 10 – sintered $NbC_{\sim 1.0}$, 0.5 N load [5, 63, 64]; 11 – sintered $NbC_{\sim 1.0}$, porosity – 28%, 1.5 kN load [295] (*Inset* – hardness of single crystal $NbC_{0.87}$ (001) produced by floating zone technique, in lg(*HV*, GPa) – 1/*T*, K^{-1} scale [292])

carbon content [709], and Fig. 4.24 – that of sintered materials with preceding high-temperature plastic deformation in the concentration range from $NbC_{0.74}$ to $NbC_{0.94}$ [278]. The hardness *HV*, GPa at room temperature of stoichiometric niobium semicarbide Nb_2C produced by gas carburization is 20.5 [53]; for sintered materials with the same composition it is 20.4 ± 0.5 [2, 51, 67] or 20.9 [10, 11, 289]. The data on microhardness *Hμ* (hardness *HK*) of niobium monocarbide single crystals and various compositions of niobium semicarbide and monocarbide phases are listed in Tables 4.10–4.11. The hardness of niobium carbide thin films varies considerably, first of all depending on manufacturing methods and compositions: for filtered vacuum cathodic arc deposited films on silicon (thickness 0.5–0.6 μm) a maximum hardness is ∼45 GPa (5 mN load) [305], for electron beam deposited

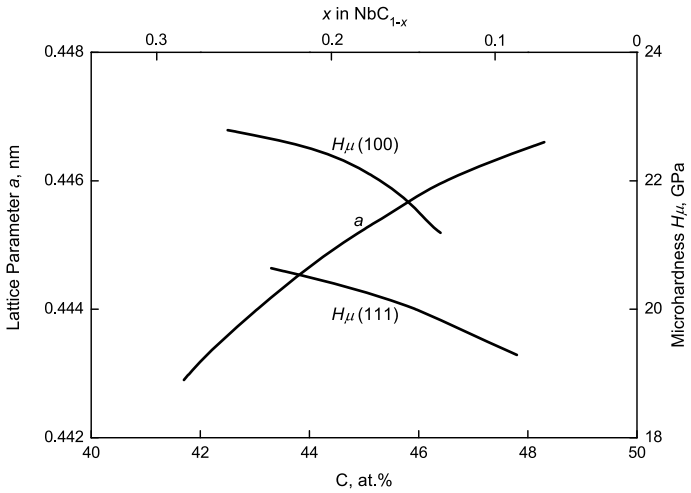

Fig. 4.23 Variations of lattice parameter a and microhardness $H\mu$ on the surfaces (100) and (111) of single crystal NbC_{1-x} with carbon content [709]

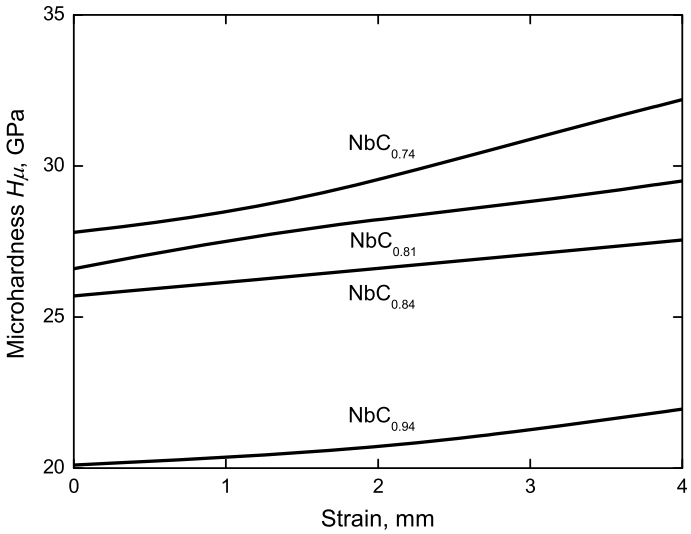

Fig. 4.24 Variations of microhardness $H\mu$ (0.5–1.0 N load) at room temperature with preceding high-temperature plastic deformation (strain – bending deflection at 1800–2230 °C, loading time 30 min) for the produced by sintering in vacuum NbC_{1-x} materials (porosity – 2–3%), having different deviations from the stoichiometry [278]

Table 4.10 Microhardness $H\mu$ (HK scale) of niobium monocarbide NbC_{1-x} ($0 < x \leq 0.25$) single crystals at room temperature (1 N load)

Surface	Indentor diagonal direction		Microhardness $H\mu$, GPa	Reference
		$x \approx 0$[a]		
(100)	<100>		23.0	[1, 274]
	<110>		23.0	[1, 274]
	<210>		23.0	[1, 274]
(001)	<100>		22.5	[1049, 1050]
	<110>		23.7	[1049, 1050]
		–	23.0[b,c]	[1, 273]
(110)	<100>		22.0	[1, 274]
	<110>		22.0	[1, 274]
	<111>		23.0	[1, 274]
	<112>		23.0	[1, 274]
(111)	<110>		21.0	[1, 274]
	<112>		21.0	[1, 274]
	<123>		21.0	[1, 274]
		$x = 0.025$[a]		
(100)	<110>		23.0	[1, 274]
	<210>		23.0	[1, 274]
(110)	<100>		23.0	[1, 274]
	<110>		22.0	[1, 274]
	<111>		23.0	[1, 274]
	<112>		23.0	[1, 274]
(111)	<100>		22.0	[1, 274]
	<110>		21.0	[1, 274]
	<112>		21.0	[1, 274]
	<123>		21.0	[1, 274]
		$x = 0.03$–0.15[a]		
(100)		–	23.1–25.2	[707]
		$x = 0.06$		
(001)		–	27.9 ± 0.5	[1, 275]
		–	27.0 ± 0.5[d]	[1, 275]
(110)		–	24.0 ± 0.5	[1, 275]
		–	20.8 ± 0.5[d]	[1, 275]
(111)		–	23.8 ± 0.5	[1, 275]
		–	19.9 ± 0.5[d]	[1, 275]
		$x = 0.08$[e]		
(111)		–	19.8	[1, 276]
		$x = 0.10$[f,g]		
(100)	<001>		~23	[284]
(110)	<001>		~24	[284]
(111)	<110>		~27	[284]
		$x = 0.13$		
(100)	<100>		17.6 ± 0.8[h]	[703]
	<110>		20.0 ± 0.8[h]	[703]
	<010>		17.6 ± 0.8[h]	[703]
		–	22.2[f]	[292]

(continued)

Table 4.10 (continued)

Surface	Indentor diagonal direction		Microhardness $H\mu$, GPa	Reference
		x = 0.14		
(100)		–	22.0[e]	[1, 276]
		–	21.6 ± 1.8[b]	[1, 268]
(102)		–	20.1 ± 1.7[b]	[1, 268]
(111)		–	20.3[e]	[1, 276]
		x = 0.18		
(100)	<100>		21.2 ± 1.4[b,i]	[87, 179]
	<110>		17.7 ± 1.4[b]	[87]
		–	22.4[e]	[1, 276]
		x = 0.19		
(100)	<100>		21.7 ± 0.9[b,j]	[2, 87]
(111)		–	20.5[e]	[1, 276]
		x = 0.20		
(100)	<100>		22.4 ± 0.9[b,j]	[2, 87]
	<001>		29.9 ± 0.6[f]	[283]
	<010>		29.5 ± 0.6[f]	[283]
	<100>		29.9[g]	[406]
	<011>		28.3 ± 1.2[f]	[283]
	<110>		27.9[g]	[406]
		–	17.5[b,k]	[87, 285]
		x = 0.21[e]		
(100)		–	22.8	[1, 276]
		x = 0.22[b]		
(100)	<100>		21.4 ± 0.7[l]	[2, 87]
		x = 0.23[e]		
(100)		–	23.0	[1, 276]
(111)		–	21.0	[1, 276]
		x = 0.25[e]		
(100)		–	23.2	[1, 276]

[a]Crystals produced by induction zone melting, contents: non-combined C – 0.05–0.50%, O – 0.05%, N – 0.005%, W – 0.22–0.50%
[b]Crystals produced by plasma-arc melting
[c]26.5 GPa at −196 °C
[d]Annealed at 2100 °C
[e]Crystals produced by arc melting
[f]Crystals produced by radio-frequency floating zone process in Ar atmosphere
[g]2 N load
[h]Crystals produced by diffusion saturation, *HV* scale, 2 N load
[i]*HV* scale (at 5 N load the measured value is 17.9 ± 0.3 GPa)
[j]*HV* scale (at 5 N load the measured value is 18.3 ± 0.3 GPa)
[k]10 N load
[l]*HV* scale (at 5 N load the measured value is 18.1 ± 0.5 GPa)

films $NbC_{\sim 1.0}$ on titanium (thickness $\sim$0.2 μm) intrinsic hardness *HV* is (18–21) ± 1 GPa (0.1–20 N load) [306], for nanocrystalline films deposited by nonreactive unbalanced DC-magnetron sputtering (thickness 0.5–0.7 μm, mean grain size 12 nm) a maximum hardness is 23 GPa (corresponds to 50 at.% C content) [97] and for films deposited using reactive DC-magnetron sputtering in discharging a mixture of CH_4 and Ar it is $\sim$37 GPa [98].

Table 4.11 Microhardness $H\mu$ of the various compositions of niobium carbide phases at room temperature

Composition	Microhardness $H\mu$, GPa	Remarks	Reference
$Nb_{2.38}C$	18.5 ± 1.1	0.5 N load, sintered in vacuum	[1, 51, 67]
$Nb_{2.22}C$	19.2 ± 0.8	0.5 N load, sintered in vacuum	[1, 51, 67]
$Nb_{2.13}C$	19.6 ± 0.6	0.5 N load, sintered in vacuum	[1, 51, 67]
$Nb_{2.04}C$	20.2 ± 0.6	0.5 N load, sintered in vacuum	[1, 51, 67]
$Nb_{2.00}C$	20.4 ± 0.5	0.5 N load, sintered in vacuum	[1, 51, 67]
$Nb_{\sim 2.0}C$	26.8	DFT-estimated value	[1097]
	25.8 ± 4.1	Produced by metal carbidization in vacuum	[288]
	24.6 ± 3.1	8 mN load, angular inclusions in U-Nb alloys	[1098]
	20.8 ± 2.0	–	[63, 82]
	20.6	–	[195]
$NbC_{0.70}$	27.0 ± 1.2	1.0 N load, melted in Ar and annealed in vacuum at 2000 °C for 4 h	[1, 277]
$NbC_{0.72}$	23.9 ± 0.6	0.5 N load, hot-pressed and annealed in vacuum, porosity – 6–10%, contents: non-combined C – 0.11%	[67, 191]
	20.1	1.0 N load, sintered at 2500–2700 °C, porosity – 10%, contents: non-combined C $\leq$ 0.01%, O $\leq$ 0.05%	[2, 282]
$NbC_{0.73}$	18.8	0.5 N load, hot-pressed and annealed in vacuum, content non-combined C – 0.15%	[848]
$NbC_{0.74}$	27.8	0.5–1.0 N load, sintered in vacuum, porosity – 2%	[1, 278]
$NbC_{0.75}$	21.9[a]	1.0 N load, sintered in vacuum and annealed, mean grain size – 29 μm, contents: non-combined C – 0.05%, N – 0.27%	[1]
$NbC_{0.76}$	29.1	The maximal value on cross-section of stock materials	[735, 736, 907]
	25.1 ± 0.8	0.5 N load, hot-pressed and annealed in vacuum, porosity – 6–10%, contents: non-combined C – 0.13–0.19%, O – 0.006%, N – 0.05%	[1, 67, 191]
$NbC_{0.80}$	28.0 ± 1.0	1.0 N load, melted in Ar and annealed in vacuum at 2000 °C	[1, 277]

(continued)

Table 4.11 (continued)

Composition	Microhardness $H\mu$, GPa	Remarks	Reference
$NbC_{0.81}$	26.6	0.5–1.0 N load, sintered in vacuum, porosity – 2%	[278]
	22.0	1.0 N load, sintered in vacuum and annealed, porosity – 12%, mean grain size – 14 μm, contents: non-combined C – 0.05%, O – 0.07%, N – 0.38%	[1]
	20.0	0.5 N load, hot-pressed and annealed in vacuum, content non-combined C – 0.23%	[848]
$NbC_{0.82}$	25.2 ± 1.0	0.5 N load, hot-pressed and annealed in vacuum, porosity – 6–10%, contents: non-combined C – 0.15–0.18%, O – 0.10%, N – 0.03%	[1, 67, 191, 209]
$NbC_{0.83}$	25.5	Produced by carbon diffusion saturation	[2, 281]
$NbC_{0.84}$	25.7	0.5–1.0 N load, sintered in vacuum, porosity – 3%	[278]
	24.0[b]	1.0 N load, sintered in vacuum and annealed, porosity – 11%, mean grain size – 18 μm, contents: non-combined C – 0.05%, O – 0.08%, N – 0.39%	[1]
$NbC_{0.85}$	26.5 ± 1.2	1.0 N load, melted in Ar and annealed in vacuum at 1800 °C	[1, 277]
$NbC_{0.86}$	21.4	1.0 N load, sintered at 2500–2700 °C, porosity – 12.5%, contents: non-combined C ≤ 0.01%, O ≤ 0.05%	[2, 282]
$NbC_{0.87}$	21.0 ± 2.2	0.2–1.5 N load, hot-pressed and annealed at 1800 °C for 3 h, porosity – 18%, content: non-combined C – 0.09%	[289, 290]
$NbC_{0.88}$	24.5	0.5 N load, hot-pressed and annealed in vacuum, content non-combined C – 0.19%	[848]
$NbC_{0.89}$	19.2 ± 0.9	Produced by carbon diffusion saturation	[82, 281]
$NbC_{0.90}$	23.0 ± 1.2	1.0 N load, melted in Ar and annealed in vacuum at 1800 °C	[1, 277]
$NbC_{0.91}$	26.5	0.5 N load, hot-pressed and annealed in vacuum, content non-combined C – 0.52%	[848]
	24.8 ± 1.0	0.5 N load, hot-pressed and annealed in vacuum, porosity – 6–10%	[209]
	22.0	1.0 N load, sintered at 2500–2700 °C, porosity – 2%, contents: non-combined C ≤ 0.01%, O ≤ 0.05%	[2, 282]
$NbC_{0.92}$	23.6 ± 0.9	0.5 N load, hot-pressed and annealed in vacuum, porosity – 6–10%, contents: non-combined C – 0.18%, O – 0.10%, N – 0.03%	[1, 67, 191]
	18.0[c]	1.0 N load, sintered in vacuum and annealed, porosity – 7%, mean grain size – 12 μm, contents: non-combined C – 0.05%, O – 0.07%, N – 0.32%	[1]
$NbC_{0.93}$	15.7	Produced by carbon diffusion saturation	[2, 281]
$NbC_{0.94}$	20.1	0.5–1.0 N load, sintered in vacuum, porosity – 3%	[1, 278]
$NbC_{0.95}$	21.1	1.0 N load, sintered at 2400 °C, porosity ~10–30%, contents: non-combined C – 0.35%, O – 0.05%, N – 0.005%, W – 0.22%	[707]
	20.0	1.0 N load, sintered at 2500–2700 °C, porosity – 6%, contents: non-combined C ≤ 0.01%, O ≤ 0.05%	[2, 282]

(continued)

Table 4.11 (continued)

Composition	Microhardness $H\mu$, GPa	Remarks	Reference
$NbC_{0.96}$	16.7 ± 1.0	1.0 N load, sintered at 2500–2700 °C, porosity – 10%, contents: non-combined C ≤ 0.01%, O ≤ 0.05%	[2, 282]
$NbC_{0.97}$	18.6 ± 1.0	1.0 N load, produced by carbon diffusion saturation (carbidization) of pure metal, contents: O – 0.05%, N – 0.06%	[1, 2, 279]
	14.3 ± 1.0	Produced by carbon diffusion saturation	[2, 281]
$NbC_{0.98}$	21.2	Deposited from gas phase	[1, 280]
	20.0 ± 1.2	1.0 N load, melted in Ar and annealed in vacuum at 1800 °C	[1, 277]
	20.0 ± 1.0	0.5 N load, hot-pressed and annealed in vacuum, porosity – 6–10%	[209]
	13.8 ± 0.8	Pyrolytic materials	[779]
$NbC_{0.99}$	23.5	Sintered materials	[769]
	21.3 ± 1.3	0.5 N load, hot-pressed and annealed in vacuum, porosity – 6–10%, contents: non-combined C – 0.25%, O – 0.03%, N – 0.01%	[1, 67, 191]
	15.0[d]	1.0 N load, sintered in vacuum and annealed, porosity – 7%, mean grain size – 12 μm, contents: non-combined C – 0.07%, O – 0.05%, N – 0.45%	[1]
$NbC_{\sim 1.0}$	24.5	Recommended value	[295, 778]
	24.2[e]	The highest value for materials sintered in vacuum (with fugitive binder), porosity – 1–5%	[177, 333]
	24.1	Derived from first principles calculations	[154]
	24.0	–	[7, 722, 724]
	23.5	0.5–1.0 N load	[104, 177, 189, 286]
	23.0[f]	Nanocrystalline coatings deposited by dual DC magnetron sputtering, thickness – 0.5–0.7 μm, ~15% amorphous carbon phase	[97]
	~23[e,g]	1.0 N load, arc-cast	[294]
	22.7 ± 3.3	Produced by metal carbidization in vacuum	[288]
	22.5	Calculated theoretically on the basis of Gao's model	[301, 780]
	21.9	First-principles calculated	[955]
	(21.0–21.6) ± (1.2–2.2)	–	[10, 287]
	20.1	–	[1051]
	20.0	–	[64, 726, 905]
	19.6	Averaged on the basis of several sources	[84, 110, 291, 293]
	19.2	0.3 N load	[65, 170]
	19.1[h]	In contact with graphite phase	[773]
	18.7	Calculated theoretically on the basis of Šimůnek model	[301, 781]

(continued)

Table 4.11 (continued)

Composition	Microhardness $H\mu$, GPa	Remarks	Reference
	18.6[i]	0.6–0.8 N load, hot-pressed, porosity – 7%	[343]
	(18.0–21.0) ± 1.0	0.1–19.6 N load, electron beam deposited thin films on Ti substrate, thickness – 0.2 μm	[306]
	18.0	–	[178, 299]
	18.3	Calculated theoretically on the basis of Šimůnek-Vackář's model	[302, 304]
	17.6–19.6	–	[301, 302]
	17.5 ± 0.8	Produced by metal fibre carbidization	[1]
	16.8 ± 0.9	2.0 N load, hot-pressed, porosity <2%	[322, 323]
	16.1	Calculated theoretically on the basis of Tian's model	[301–303]
	14.9[j]	0.2–10 N load, hot-pressed, porosity <5%	[702]
	12.7 (?)	–	[725]

[a]3.9 GPa – at 1000 °C
[b]4.3 GPa – at 1000 °C
[c]3.7 GPa – at 1000 °C
[d]3.5 GPa – at 1000 °C
[e]Knoop measurement (*HK*)
[f]Nanoindentation with a depth set to 50 nm and loading rate – 3mN min^{-1}
[g]For 10 N load with boron carbide indentor: 4.0 GPa – at 1000 °C and 0.8 GPa – at 1800 °C
[h]Diamond pyramid hardness measurement (*DPH*)
[i]5.3 GPa at – 1000 °C and 2.0 GPa – at 1700 °C
[j]Determined in accordance with the formula: $H\mu = 18.18P/d^2$, where P is the load on the indentor, N and d is the diagonal of the indentation, μm; the values of other micromechanical characteristics: microbrittleness $\gamma_\mu = (D^2 - d^2)/d^2 = 0.75$, where D is the average size of the damageability zone, μm and brittle microstrength $\sigma_\mu = P/D^2 = 4.6$ GPa [702]

On the basis of several sources the variations of ultimate tensile σ_t, flexural (bending) σ_f and compressive σ_c strengths of niobium monocarbide NbC_{1-x} materials with temperature and carbon content within the homogeneity range are shown in Figs. 4.25–4.28. The compressive strength of $NbC_{\sim 1.0}$ crystallites is at least 24 ± 4 GPa at a confining pressure of 57 ± 0.5 GPa [379]. The approximate relationship of $\sigma_t : \sigma_f : \sigma_c = 1 : (1.5–2.0) : (8–10)$ can be recommended for the general estimations of the strength characteristics of NbC_{1-x} materials [727, 732, 736]. At ambient temperatures niobium monocarbide NbC_{1-x} materials exhibit fracture toughness K_{IC} of 6.0–8.6 MPa $m^{1/2}$ (for hot-pressed nanostructured materials $NbC_{\sim 1.0}$, porosity – 2%, mean grain size – 76 nm) [321, 902], 4.0 MPa $m^{1/2}$ (for sintered $NbC_{0.96}$, contents: O $\leq$ 0.05%, N – 0.05%, W – 0.2–0.3%) [731] and 3.5–3.8 (for $NbC_{\sim 1.0}$ single crystals) [1, 2, 273, 317]; apparently fracture toughness of single crystals decreases with carbon content decreasing within the homogeneity range as for single crystal $NbC_{0.80}$ (001) it is 2.8 MPa $m^{1/2}$ [1, 2, 87, 285]. Toughness K_{IC} declines with the temperature growth to

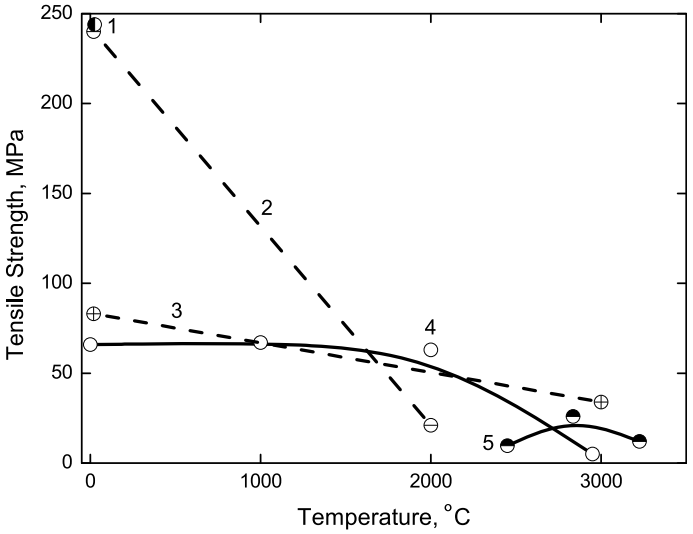

Fig. 4.25 Variation of tensile strength of near-stoichiometric niobium monocarbide NbC_{1-x} with temperature on the basis of several sources: 1 – [338, 340]; 2 – [170, 311]; 3 – [10, 11]; 4 – sintered, 30% porosity [1, 2, 67, 309]; 5 – sintered, corrected to the poreless state [310] (the value of first-principles calculated ideal tensile strength for quasi-stoichiometric NbC_{1-x} is 72.5 GPa [955])

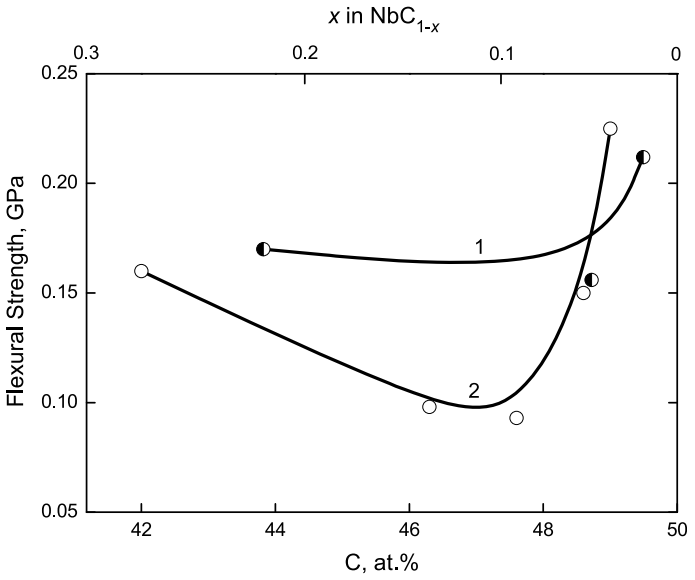

Fig. 4.26 Variation of the flexural (bending) strength at room temperature of sintered niobium monocarbide NbC_{1-x} materials with deviation from the stoichiometry within the homogeneity range: 1 – porosity 6–24%, mean grain size 8–23 μm, content non-combined C 0.05–1.0% [315]; 2 – porosity 2–13%, content O < 0.05% [2, 282]

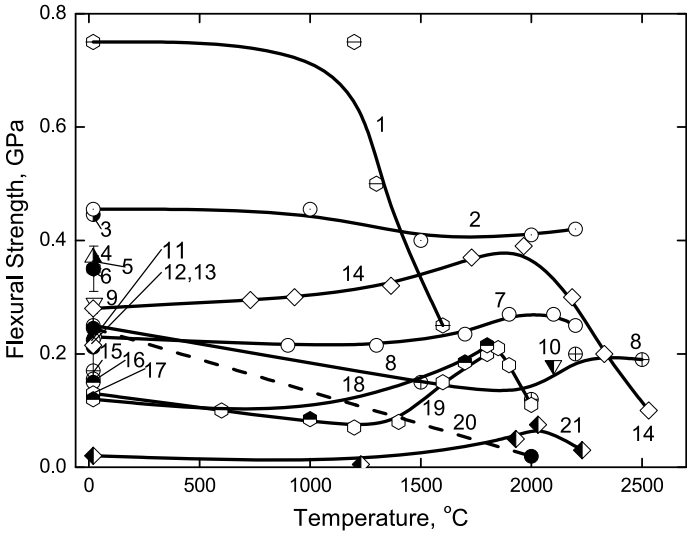

Fig. 4.27 Variation of flexural (bending) strength of near-stoichiometric niobium monocarbide NbC_{1-x} materials with temperature on the basis of several sources: 1 – $NbC_{0.84}$, single crystal materials produced by plasma-arc melting and cut off along the <100> axis, 3-point bending scheme, measured in vacuum ($\leq 1.3 \times 10^{-2}$ Pa) [1, 268, 273]; 2 – $NbC_{0.97}$ produced by metal right-through saturation in Ar atmosphere, cylinder specimen with 0.6 mm diameter, 3-point bending scheme [1, 313]; 3 – $NbC_{0.97}$ produced by metal carbidization (saturation), mean grain size – 350–450 μm, contents: O + N – 0.1% [2, 279]; 4 – [336]; 5 – sintered materials, porosity – 3–5% [1, 314]; 6 – hot-pressed materials, porosity < 2%, 4-point bending scheme (Weibull modulus $m = 9$) [322, 323]; 7 – sintered materials, porosity – 8–10% [2, 312]; 8 – $NbC_{0.97}$ produced by metal right-through saturation in Ar atmosphere, 0.6 × 20 × 25 mm plate, 3-point bending scheme [1, 313]; 9 – porosity 7–15%, 3-point bending scheme [200, 338]; 10 – [335]; 11 – sintered $NbC_{0.96}$, porosity – 6%, content O < 0.05% [2, 282]; 12 – sintered $NbC_{0.98}$, porosity – 6%, mean grain size – 8 μm, contents: non-combined C – 1% [2, 315]; 13 – sintered materials, porosity – 8%, mean grain size – 15 μm, cylinder 3 mm diameter, 4-point bending scheme with 80 mm base (Weibull modulus $m = 3.8$) [727]; 14 – sintered materials [735, 736]; 15 – sintered $NbC_{0.76}O_{0.02}N_{0.01}$, porosity – 20%, mean grain size – 23 μm, content non-combined C – 0.05% [2, 315]; 16 – sintered $NbC_{0.96}O_{0.02}$, porosity – 24%, mean grain size – 17 μm, content non-combined C – 0.24% [2, 315]; 17 – sintered $NbC_{0.945}$, porosity – 5%, content: O < 0.05% [2, 282]; 18 – sintered materials, annealed at 2000 °C for 10 min [1, 316]; 19 – sintered materials, non-annealed [1, 316]; 20 – sintered materials, porosity – 4% [177]; 21 – hot-pressed (low-porous) materials [785, 786] (the values of first-principles calculated ideal tensile and shear strengths for quasi-stoichiometric NbC_{1-x} are 72.5 and 82 GPa, respectively [955])

3.2 MPa $m^{1/2}$ at 500 °C [273] and up to 1.8–2.0 MPa $m^{1/2}$ at 900 °C [1, 2]. The first-principles calculated ideal fracture toughness K_{IC} of quasi-stoichiometric NbC_{1-x} is ~5.0 MPa $m^{1/2}$ [955]. The reported values for ductile-to-brittle transition temperatures of NbC_{1-x} phases with various deviations from the stoichiometry are listed in Table 4.12. The values of prolonged strength are very important for the evaluation of operational time of components at high and ultra-high temperatures. The tensile and flexural (bending) steady-state creep isoterms of sintered niobium

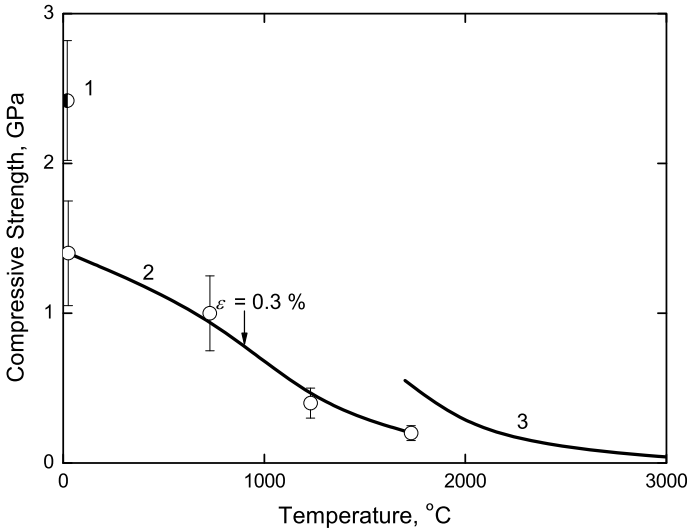

Fig. 4.28 Variation of compressive strength of near-stoichiometric niobium monocarbide NbC_{1-x} materials with temperature: 1 – low-porous $NbC_{\sim 1.0}$ [1, 2, 10, 11, 63, 67, 170, 338], 2 – $NbC_{\sim 1.0}$ [5], 3 – $NbC_{0.94}$ [10, 11] (approximate compression fracture deformation ε is indicated at 900 °C)

Table 4.12 Ductile-to-brittle transition temperatures of niobium monocarbide phases NbC_{1-x}[a,b]

Composition	Temperature, K (°C)	Characteristics	Reference
$NbC_{\sim 1.0}$	1470 (1200)	Single crystals[c]	[273]
$NbC_{0.95}$	1750 (1480)	Hot-pressed and annealed (porosity ~7.5%)	[2, 318]
$NbC_{0.84}$	1470 (1200)	Single crystals	[1, 268]
$NbC_{0.77}$	1270–1320 (1000–1050)	Single crystals	[320]
$NbC_{0.76}$	1270–1370 (1000–1100)	Single crystals[d]	[319, 410]
	1470 (1200)	Single crystals	[2, 268]

[a]For near-stoichiometric NbC_{1-x} the following average values were estimated: Peierls stress τ_P = 6.3 GPa, shear modulus – Peierls stress ratio τ_P/G = 0.037 and slip activation energy barrier q = 1.5 eV [739]

[b]According to Lanin [727, 732], with increasing porosity from 7 to 60% the ductile-to-brittle transition temperatures of NbC_{1-x} materials enlarge by 400 K at flexural (bending) testing and by 600–700 K at compression testing

[c]Based on flexural (bending) strength testing

[d]Based on compression strength testing

monocarbide NbC_{1-x} materials within the homogeneity range are given in Fig. 4.29, the examples of intermediate-, high- and ultra-high temperature steady-state creep rate $d\varepsilon/dt$ variations for near-stoichiometric niobium monocarbide NbC_{1-x} materials with applied stress and temperature are demonstrated in Figs. 4.30 and 4.31; steady-state creep activation energies Q and exponent constants n (*see* Eq. (I-3.17)) for various NbC_{1-x} phases are listed in Table 4.13. Approximately, the following main dominating creep mechanisms can be established in some overlapping parametric ranges for niobium monocarbide NbC_{1-x} materials with certain changes within the homogeneity range [328, 330, 787–789]:

vacancy-diffusional (or Nabarro-Hering) mechanism controlled by the bulk diffusion of metal in carbide (at temperatures from 2300 to 3000 °C and applied stresses from 7 to 30 MPa, with $Q \approx 450$–550 kJ mol^{-1} and $n = 1$);
dislocation-diffusional mechanism controlled by the bulk diffusion of metal in carbide (at 2300–3000 °C and 30–70 MPa, with $Q \approx 450$–550 kJ mol^{-1} and $n \approx 3$–4);
viscous dislocation sliding mechanism controlled by the bulk diffusion of carbon in carbide (at 1800–2200 °C and 50–200 MPa, with $Q \approx 330$–350 kJ mol^{-1} and $n \approx 2.5$–3.5).

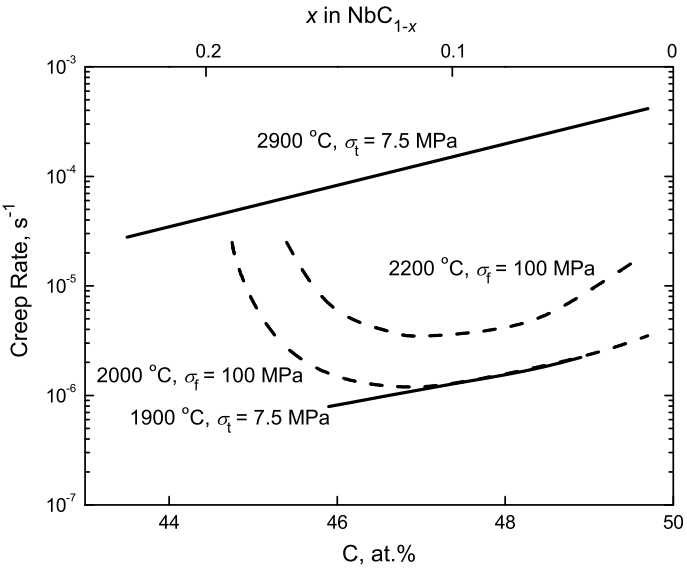

Fig. 4.29 Tensile and flexural (bending) steady-state creep isoterms of sintered niobium monocarbide NbC_{1-x} materials with deviation from stoichiometry within the homogeneity range at various temperatures and stresses [1, 3, 269, 281, 324, 330]

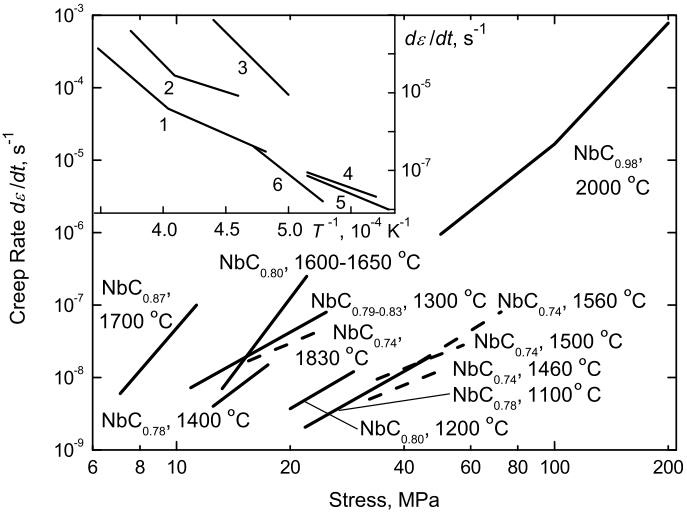

Fig. 4.30 Intermediate- and high-temperature steady-state creep rate $d\varepsilon/dt$ variations of near-stoichiometric niobium monocarbide NbC_{1-x} materials with applied stress on the basis of several sources: $NbC_{0.74}$ isostatically hot-pressed, porosity – ~0.5%, mean grain size – 25 μm, under compressive load at 1460, 1500, 1560 and 1830 °C [342]; $NbC_{0.78}$, $NbC_{0.79–0.83}$, $NbC_{0.80}$ and $NbC_{0.87}$ single crystals produced by floating zone technique, contents: Ta – ~0.08%, other impurities (in total) – ~0.005%, density of dislocations – ~2.6×10^9 m^{-2}, under compressive load at 1100, 1200, 1300, 1400, 1600–1650 and 1700 °C [320]; $NbC_{0.98}$ sintered, porosity – 2%, mean grain size – 20 μm, contents: non-combined C – 0.08%, O – 0.05%, N – 0.01%, W – 0.55%, under compressive load at 2000 °C [327, 328] (*Inset* – variations with temperature: $NbC_{0.98}$ sintered, porosity – 2%, mean grain size – 20 μm, contents: non-combined C – 0.08%, O – 0.05%, N – 0.01%, W – 0.55%, under compressive load of: 1 – 60, 2 – 100 and 3 – 200 MPa [327, 328] and $NbC_{0.74}$ isostatically hot-pressed, porosity – ~0.5%, mean grain size – 25 μm, under compressive load of: 4 – 54, 5 – 48 and 6 – 64 MPa [342]; in $d\varepsilon/dt$, s^{-1} – $1/T$, K^{-1} scale)

At room temperature the values of Young's E, Coulomb's (shear) G and bulk (compression) K moduli, volume compressibility κ and Poisson's ratio ν of sub- and near-stoichiometric niobium monocarbide NbC_{1-x} materials lie within the areas: E = 280–580 GPa [1–3, 28, 84, 193, 282, 338, 346, 350–352, 741], G = 140–250 GPa [2, 3, 84, 193, 282, 338, 346–348, 351, 352, 893], K = 300–380 GPa [2, 3, 193, 338, 346, 348, 349, 351], κ = 3.26–3.38 TPa^{-1} [1–3, 353] and ν = 0.21–0.24 [2, 3, 193, 282, 338, 348, 351, 352] (*see also* Table 4.14); while for near-stoichiometric niobium semicarbide $Nb_{2\pm x}C$ the estimated value of E is 410 GPa [1, 345], the mean elastic modulus E of angular inclusions of Nb_2C in U-Nb alloys measured by nanoindentation was 247 ± 24 GPa [1098], the DFT-calculated moduli for α-Nb_2C (*Pnma*) are E = 307 GPa, K = 226 GPa, G = 121 GPa and ν = 0.27 [154], or E = 427 GPa, K = 236 GPa, G = 178 GPa and ν = 0.19 [1097], for α-Nb_2C (*Pm*) the values are E = 221 GPa, G = 84 GPa, K = 203 GPa and ν = 0.31 [405], for β-$Nb_{2+x}C$ (*P(–3)m1*): E = 310 GPa, K = 229 GPa, G = 122 GPa and ν = 0.27 [405], and the values of K estimated by Abderrahim et al. [358] for α-, β- and γ-modifications of niobium semicarbide are

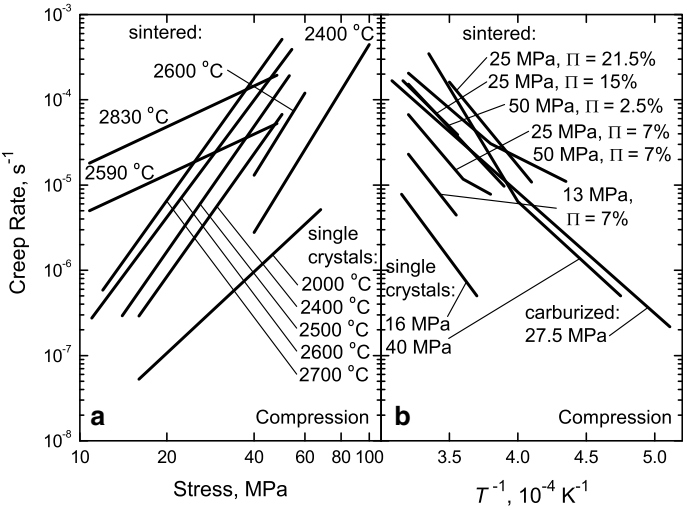

Fig. 4.31 High- and ultra-high temperature steady-state creep rate variations of near-stoichiometric niobium monocarbide NbC_{1-x} materials on the basis of several sources: (**a**) – with applied stress ($NbC_{0.90–0.94}$ sintered in vacuum, porosity – 7%, mean grain size – 15 μm, rate after 10 min testing under compressive load at 2590 and 2830 °C [329]; $NbC_{0.98}$ sintered, porosity – 2%, mean grain size – 20 μm, contents: non-combined C – 0.08%, O – 0.05%, N – 0.01%, W – 0.55%, under compressive load at 2400 and 2600 °C [327, 328]; $NbC_{\sim 1.0}$ single crystal materials, steady-state creep along the <110> direction under compressive load at 2000, 2400, 2500, 2600 and 2700 °C [326]) and (**b**) – with temperature ($NbC_{0.90–0.94}$ sintered in vacuum with porosity – 2.5–21.5% and mean grain size – 8–36 μm, rate after 10–30 min testing under compressive load of 13, 25 and 50 MPa [329, 332]; $NbC_{\sim 1.0}$ single crystal materials, steady-state creep along the <110> direction under compressive load of 16 and 40 MPa [326]; $NbC_{\sim 1.0}$ metal-carburized materials under compressive load of 27.5 MPa [338, 341])

Table 4.13 Formal creep characteristics (activation energy Q, stress exponent constant n) of niobium monocarbide phases NbC_{1-x} at various temperature and applied stress ranges

Composition	Load type[a]	Temperature range, °C	Applied stress range, MPa	Activation energy[b], Q, kJ mol^{-1}	Exponent constant[b], n	Reference
$NbC_{\sim 1.0}$	–	Low temp.	–	~360	3.0	[893]
	–	High temp.	–	~500	4.8	[893]
$NbC^{c}_{0.89–0.98}$	T, F, C[d]	<2350	–	330–345	2.8–3.2	[324, 325]
$NbC^{c}_{0.925}$	T, F, C[d]	>2350	$\sigma < \sigma^{e}_{th}$	525	1[f]	[324, 325]
	T, F, C[d]	>2350	$\sigma > \sigma^{e}_{th}$	525	3.4–3.7	[324, 325]
$NbC^{g}_{0.98}$	C	1700–2200	60–100	240	3.7	[327, 328]
	C	1700–2000	200	~600	5.4	[327, 328]
	C	2200–2600	60–200	~600	5.4	[327, 328]
$NbC^{h}_{0.97}$	T	2500–2900	1.5–8.5	450 ± 20	1	[2, 281]
$NbC^{h}_{0.93}$	T	2500–2900	1.5–8.5	450 ± 20	1	[2, 281]
$NbC^{h}_{0.89}$	T	2500–2900	1.5–8.5	590 ± 30	1	[2, 281]
$NbC^{h}_{0.83}$	T	2500–2900	1.5–8.5	420	1	[2, 281]
$NbC_{0.96}$	T	2600-2900	–	565	1	[281, 344]
$NbC^{i}_{0.90–0.94}$	C	2430–2930	51	345 ± 15	1.66	[329]
	C	2430–2930	25	390 ± 15	1.66	[329]
	C	2430–2930	13	500 ± 15	1.66	[329]

(continued)

Table 4.13 (continued)

Composition	Load type[a]	Temperature range, °C	Applied stress range, MPa	Activation energy[b], Q, kJ mol^{-1}	Exponent constant[b], n	Reference
$NbC^{j}_{0.92–0.93}$	C	2300–3000	7–40	520	1	[2, 330]
	C	2300–3000	40–70	520	3.7	[2, 330]
	F	1800–2200	50–200	340	2.9	[330]
$NbC^{k}_{0.92}$	C	2550–3100	20–100	360 ± 20	–	[331, 332]
$NbC^{l}_{0.89}$	C	1800–2200	12–80	335 ± 35	3.3 ± 0.1	[326, 330]
	C	2200–2700	12–80	500 ± 40	4.8 ± 0.3	[326, 330]
$NbC^{m}_{0.78–0.80}$	C	1100–1200	20–40	–	3.0–3.2	[320, 710]
$NbC^{m}_{0.78–0.83}$	C	1300–1400	10–25	290	3.0–4.5	[320, 710]
$NbC^{m}_{0.78–0.87}$	C	1400–1700	7–25	∼550	6.2–6.8	[320, 710]
$NbC^{n}_{0.74}$	C	1450–1830	16–55	230–250	2	[342, 691]
	C	1560–1830	55–70	470	3.2	[342, 691]
$NbC^{o}_{0.90}$	H	1000–1500	–	320–420	3.2–4.7	[283, 284]
$NbC^{p}_{0.80}$	H	1000–1500	–	380	3.9	[283, 284]
$NbC^{q}_{0.98}$	F	1800–2200	50–200	335	2.8	[330]
$NbC_{0.98}$	F	2465–2685	10	645	–	[657, 658]
$NbC_{0.91}$	F	2430–2775	10	580	–	[657, 658]
$NbC^{r}_{0.89}$	F	1800–2200	50–200	330	3.1	[330]
$NbC_{0.86}$	F	2305–2775	10	475	–	[657, 658]
$NbC_{0.81}$	F	2305–2775	10	515	–	[657, 658]
$NbC_{0.76}$	F	2250–2685	10	410	–	[657, 658]
$NbC_{0.71}$	F	2190–2580	10	350	–	[657, 658]

[a]Denoted: T – tension, F – flexure (bending), C – compression, H – hot hardness (measurement)
[b]For Q and n *see* Eq. (I-3.17) and consideration in [893]
[c]Porosity – 3–4%, mean grain size L = 12–20 μm
[d]No effect of the character of loading on the creep mechanism and rate
[e]σ_{th} – threshold value of stress
[f]Creep rate $d\varepsilon/dt \sim 1/L^2$
[g]Sintered, porosity – 2%, mean grain size – 20 μm, contents: non-combined C – 0.08%, O – 0.05%, N – 0.01%, W – 0.55%
[h]Produced by metal right-through saturation, mean grain size L = 130 μm
[i]Sintered in vacuum, porosity – 7–15%, mean grain size L = 7–12 μm (n = 2.3 for samples with porosity – 21.5%, mean grain size L = 8 μm and n = 2.5 for samples with porosity – 32.5%, mean grain size L = 3–4 μm)
[j]Sintered in vacuum, porosity 3–4%, mean grain size L = 20 μm
[k]Sintered in vacuum, porosity – 2%, mean grain size L = 36 μm, contaminated with 1.4% WC
[l]Single crystals produced by zone refining, contents: non-combined C ≤ 0.05%, N – 0.01%, Si – 0.03%
[m]Single crystals produced by floating zone technique, contents: Ta – ∼0.08%, other impurities (in total) – ∼0.005%, density of dislocations – $\sim 2.6 \times 10^9$ m^{-2}
[n]Isostatically hot-pressed (HIP), porosity – ∼0.5%, mean grain size L = 25 μm
[o]Single crystal prepared by floating-zone process, data summarized from plane directions (100) <001>, (110) <001> and (111) <110>
[p]Single crystal prepared by floating-zone process, plane direction – (100) <001>
[q]Sintered, porosity 3–4%, mean grain size L = 12 μm
[r]Sintered, porosity 3–4%, mean grain size L = 18 μm

Table 4.14 Elastic properties (stiffness coefficients c_{11}, c_{12} and c_{44}, Young's modulus E, bulk (compression) modulus K, Poisson's ratio ν)[a] of niobium monocarbide phases NbC_{1-x}

Composition	Stiffness coefficients c_{ij}			Young's modulus E, GPa	Bulk modulus K, GPa	Poisson's ratio v	Reference
	c_{11}, GPa	c_{12}, GPa	c_{44}, GPa				
$NbC^{b}_{0.72}$	–	–	–	330	–	–	[1, 360]
$NbC_{0.73}$	–	–	–	342[c]	189[c]	–	[2]
$NbC_{0.75}$	–	–	–	372[b,d]	–	0.21	[2, 282]
	–	–	–	365 ± 7[e]	–	–	[1, 361]
	542[f]	–	148[f]	–	–	–	[369]
	512[g]	150[g]	109[g]	345[g,h]	270[g,h]	–	[393]
	–	–	–	–	227	–	[883]
$NbC_{0.77}$	–	–	–	385[b,i]	231[i]	–	[1, 2, 360]
	–	–	–	360 (320)[j]	–	–	[1, 352]
$NbC_{0.80}$	–	–	–	363[b,k]	–	0.23	[2, 282]
	–	–	–	–	225	–	[883]
$NbC^{l}_{0.81}$	–	–	–	367 ± 7	221	0.22	[2, 361]
$NbC^{b}_{0.83}$	–	–	–	430[m]	273[m]	–	[1, 2, 360]
$NbC^{n}_{0.84}$	–	–	–	377 ± 7	298	0.245	[1, 2, 361]
$NbC_{0.86}$	–	–	–	333[b,o]	–	0.23	[2, 282]
	–	–	–	366[b]	206	0.23	[704]
	–	–	–	–	210	–	[883]
$NbC^{f}_{0.865}$	566	117	153	438[p]	267[p]	0.23	[369, 394]
$NbC^{g}_{0.875}$	608	126	153	452[q]	287[q]	–	[393]
$NbC^{b}_{0.88}$	–	–	–	368[r]	–	0.24	[2, 282]
$NbC_{0.885}$	–	–	–	540	–	–	[59]
$NbC^{b}_{0.90}$	–	–	–	466[r]	300[r]	–	[1, 2, 360]
$NbC^{s}_{0.90}$	413	111	206	–	–	–	[370]
$NbC_{0.91}$	–	–	–	450 (440)[h,j]	–	–	[1, 352]
	–	–	–	428[b]	255	0.23	[704]
	–	–	–	–	260	–	[883]
$NbC_{0.92}$	–	–	–	348 ± 7[t]	217[t]	0.23	[2, 361]
	–	–	–	435 (395)[j]	–	–	[1, 352]
$NbC_{0.95}$	–	–	–	475[b,u]	–	0.22	[2, 282]
	–	–	–	478[v]	–	–	[355]
	–	–	–	500[w]	–	–	[1, 352]
$NbC_{0.96}$	–	–	–	489[x]	303[x]	0.23	[348, 351]
	–	–	–	494[b,y]	–	0.22	[2, 282]
	–	–	–	500[b]	302	–	[1, 7, 905]
	–	–	–	370[z]	–	–	[731]
	–	–	–	–	300	–	[883]

(continued)

Table 4.14 (continued)

Composition	Stiffness coefficients c_{ij}			Young's modulus E, GPa	Bulk modulus K, GPa	Poisson's ratio ν	Reference
	c_{11}, GPa	c_{12}, GPa	c_{44}, GPa				
$NbC_{0.97}$	–	–	–	504[a1]	296[a1]	0.22	[193, 353]
	–	–	–	505 (460)[j]	–	–	[1, 352]
	–	–	–	518[a2]	301[a2]	0.21	[346, 353]
	–	–	–	495[b]	–	–	[1, 360]
	–	–	–	510 (490)[b,j]	–	0.21 (0.23)[j]	[1, 7, 357]
$NbC_{0.98}$	–	–	–	477[a5]	312[a5]	0.24	[2, 3, 360]
	–	–	–	470[a5]	305[a5]	0.24	[2, 364]
$NbC_{0.99}$	–	–	–	469 ± 8[a6]	300[a6]	0.24	[1, 2, 361]
$NbC_{\sim 1.0}$	–	–	–	421[a7]	–	–	[1, 27]
	–	–	–	569[a8]	–	–	[362, 408]
	–	–	–	338[a9]	–	–	[7, 363, 374, 724, 1054]
	–	–	–	539–540[a9]	–	0.10 (?)	[5, 346, 350]
	–	–	–	274[b1]	–	–	[170]
	–	–	–	340	–	0.22	[80, 722]
	–	–	–	333	–	–	[195]
	–	–	–	580	–	–	[299]
	–	–	–	–	320	–	[365]
	620[b2]	200[b2]	150[b2]	441	302	0.28	[366, 367, 396, 398]
	–	–	–	–	376[b3]	–	[349]
	–	–	–	–	296[b4]	–	[349]
	–	–	–	–	326[b5]	–	[371]
	–	–	–	340	–	–	[178]
	667[b6]	163[b6]	161[b6]	484[b6]	332[b6,b7]	0.26	[375, 396]
	627[b8]	159[b8]	151[b8]	–	326[b8]	–	[376]
	587[b9]	121[b9]	142[b9]	–	–	–	[376]
	627[b8]	179[b8]	220[b8]	–	328[b8]	–	[377]
	546[g]	167[g]	224[g]	–	293[g]	–	[377]
	640[c1]	180[c1]	140[c1]	–	332[c1]	–	[377, 378]
	–	–	–	419[c2,c3]	285[c2,c3]	0.19[c2]	[152]
	–	–	–	–	274 ± 3[c4]	–	[379]
	–	–	–	–	335[b8]	–	[380]
	–	–	–	–	319[c5]	–	[380]
	–	–	–	–	301[c6]	–	[381, 383]
	–	–	–	–	315[c7]	–	[380, 382]
	–	–	–	–	324[c8]	–	[382]
	–	–	–	–	294[c9]	–	[384]
	–	–	–	–	299[d1]	–	[384]
	648[d2]	109[d2]	109[d2]	617[d2,d3]	288[d2,d3]	0.16[d2]	[385, 386]
	597[g]	152[g]	167[g]	–	301[g]	–	[380, 390]

(continued)

Table 4.14 (continued)

Composition	Stiffness coefficients c_{ij}			Young's modulus E, GPa	Bulk modulus K, GPa	Poisson's ratio v	Reference
	c_{11}, GPa	c_{12}, GPa	c_{44}, GPa				
	–	–	–	–	333[d4]	–	[389]
	–	–	–	–	357[g]	–	[391]
	–	–	–	–	360[d5]	–	[391]
	–	–	–	295[d6]	–	–	[97]
	–	–	–	–	333[d7]	–	[395]
	–	–	–	–	290[d8]	–	[395]
	651[d9]	125[d9]	161[d9]	484[d9,e1]	301[d9,e1]	0.23[d9]	[396, 401]
	–	–	–	–	293[e2]	–	[397]
	617[e3]	199[e3]	159[e3]	520[e3,e4]	338[e3,e4]	0.24[e3]	[399]
	460[e5]	159[e5]	103[e5]	378[e5,e6]	259[e5,e6]	0.26[e5]	[399]
	–	–	–	–	333–340[c7,e7]	–	[400]
	–	–	–	477[e8]	–	–	[322, 323]
	630[b5,e5]	200[b5,e5]	161[b5,e5]	461[b5,e5,e9]	343[b5,e5,e9]	0.28[b5,e5]	[402, 403]
	556[e5,f1]	184[e5,f1]	140[e5,f1]	402[e5,f1,f2]	308[e5,f1,f2]	0.28[e5,f1]	[402, 403]
	604[f3]	146[f3]	179[f3]	486[f3,f4]	299[f3,f4]	0.23[f3]	[154]
	–	–	–	–	298[b5,f5]	–	[356]
	557[g]	163[g]	147[g]	484[g,f6]	294[g,f6]	0.23[g]	[404, 405]
	644[g]	129[g]	166[g]	–	302[g,f7]	–	[393]
	646[f8]	127[f8]	192[f8]	406–604[f8]	300[f8]	–	[409]
	593[f9]	140[f9]	176[f9]	365–539[f9]	291[f9]	–	[409]
	–	–	–	–	311–336[g1]	–	[409]
	–	–	–	310[g2]	–	–	[727]
	593[g3]	126[g3]	132[g3]	428[g3,g4]	281[g3,g4]	0.24[g3]	[738, 739]
	–	–	–	–	303[g5]	–	[790]
	–	–	–	–	327[b8]	–	[791]
	–	–	–	–	300[b9]	–	[791]
	656[g6]	169[g6]	163[g6]	–	338[g6]	–	[792]
	–	–	–	494[g7,g8]	311[g7,g8]	0.235[g7]	[793]
	–	–	–	–	300[g9]	–	[794]
	667[g]	117[g]	168[g]	632[g,h1]	300[g,h1]	0.15[h1]	[795]
	699[h2]	130[h2]	170[h2]	515[h2,h3]	320[h2,h3]	0.231[h2]	[796]
	–	–	–	300[h4]	–	–	[731]
	–	–	–	495[h5]	–	–	[860]
	–	–	–	–	571[h6]	–	[867]
	–	–	–	–	305–332[h7]	–	[888]
	–	–	–	392[h8]	340[h8]	–	[904]
	–	–	–	–	282[d9]	–	[910]
	650[h9]	126[h9]	173[h9]	500[h9,i1]	301[h9,i1]	0.22[h9]	[926]
	–	–	–	–	285[i2]	–	[926]
	629[i3]	129[i3]	161[i3]	475[i3,i4]	296[i3,i4]	–	[955]
	–	–	–	–	295[i5,i6]	–	[387]
	–	–	–	500	–	–	[372, 373]
	–	–	–	–	318[b8]	–	[1078]

(continued)

Table 4.14 (continued)

[a]For isotropic (or quasi-isotropic) materials [3]: $E = 2G(1 + \nu)$, $E = 3K(1-2\nu)$, $K = c_{12} + 2c_{44}/3$, $G = c_{44}$; the condition for the isotropy is given in Eq. (I-2.18)
[b]Sintered, porosity – 4–12%, corrected to the poreless state
[c]G = 142 GPa
[d]G = 153 GPa
[e]Sintered in vacuum and annealed, porosity – 20%, mean grain size – 29 μm, contents: non-combined C – 0.05%, O – 0.05%, N – 0.27%
[f]Single crystal produced by a molten zone technique, contents: O – 0.008%, N < 0.0005%, Al – 0.005%, Fe < 0.005%, W – 0.02%, Zr – 0.085%
[g]Calculated on the basis of density-functional theory (DFT) with generalized gradient approximation (GGA) using the Perdew-Burke-Ernzerhof (PBE) functional
[h]Calculated value of G = 134 GPa
[i]G = 162 GPa
[j]Sintered, corrected to the poreless state (measured values are given in brackets)
[k]G = 148 GPa
[l]Sintered in vacuum and annealed, porosity – 12%, mean grain size – 14 μm, contents: non-combined C – 0.05%, O – 0.07%, N – 0.38%; measured experimentally value of G = 149 GPa
[m]G = 177 GPa
[n]Sintered in vacuum and annealed, porosity – 11%, mean grain size – 18 μm, contents: non-combined C – 0.05%, O – 0.08%, N – 0.39%; measured experimentally value of G = 140 GPa
[o]G = 136 GPa
[p]G = 178.5 GPa
[q]Calculated value of G = 181 GPa
[r]G = 188 GPa
[s]Single crystal materials
[t]Sintered in vacuum and annealed, porosity – 27%, mean grain size – 12 μm, contents: non-combined C – 0.05%, O – 0.07%, N – 0.32%; measured experimentally value of G = 141 GPa
[u]G = 195 GPa
[v]Hot-pressed, porosity – 2%, contents: non-combined C – 0.34%, O – 0.03%, N – 0.02%; measured experimentally value of G = 197 GPa
[w]Sintered, corrected to the poreless state
[x]Hot-pressed and annealed (sintered additionally), porosity – 8%, corrected to the poreless state; G = 198 GPa and $\kappa = 3.26\ \text{TPa}^{-1}$
[y]G = 202 GPa
[z]Sintered, porosity – 4%, contents: O ≤ 0.05%, N – 0.05%, W – 0.2–0.3%
[a1]Hot-pressed, porosity – 4–8%, corrected to the poreless state; measured experimentally values of G = 206 GPa and $\kappa = 3.38\ \text{TPa}^{-1}$
[a2]Sintered in vacuum, corrected to the poreless state; measured experimentally values of G = 217 GPa and $\kappa = 3.32\ \text{TPa}^{-1}$
[a3]Hot-pressed, corrected to the poreless state; measured experimentally values of G = 181 GPa and $\kappa = 3.97\ \text{TPa}^{-1}$
[a4]Calculated value of G = 192 GPa
[a5]G = 189 GPa
[a6]Sintered in vacuum and annealed, porosity – 7%, mean grain size – 12 μm, contents: non-combined C – 0.27%, O – 0.05%, N – 0.45%; measured experimentally value of G = 189 GPa
[a7]Sintered, porosity – 2%, mean grain size – 10 μm
[a8]Metal-carburized materials
[a9]Extrapolated to zero porosity
[b1]Sintered, porosity – 18%

(continued)

Table 4.14 (continued)

[b2]Estimation from phonon dispersion curves (c_{11} is extrapolated from neutron data)
[b3]Calculated on the basis of high pressure experiments by Champion and Drickamer [368]
[b4]Calculated on the basis of thermal experiments
[b5]At 0 K
[b6]Estimated theoretically by means of density-functional perturbation theory (DFPT) calculations
[b7]G = 161 GPa
[b8]Calculated on the basis of DFT with local density approximation (LDA)
[b9]Calculated on the basis DFT with GGA
[c1]Calculated on the basis of DFT with LDA using linearized augmented-plane-wave (LAPW) method
[c2]Calculated using the Debye–Grüneisen model combined with *ab initio* calculations
[c3]Calculated value of G = 167 GPa
[c4]Derived from the compression data under quasi-hydrostatic high pressure conditions
[c5]Calculated on the basis of DFT with Wu-Cohen GGA exchange-energy functional
[c6]Calculated on the basis of DFPT
[c7]Summarized value on the basis of critical review of available experimental data
[c8]Calculated on the basis of linearized method of "muffin-tin" orbitals (LMTO)
[c9]Calculated on the basis of linear combination of atomic orbitals (LCAO) method using Perdew-Wang (PW) potential
[d1]Calculated on the basis of LCAO method using Perdew potential
[d2]Calculated on the basis of interionic potential theory with modified ionic charge
[d3]Calculated value of G = 173 GPa
[d4]Calculated on the basis of DFT with LDA using full-potential linearized augmented-plane-wave (FP-LAPW) method and PW potential
[d5]Calculated on the basis of DFT with LDA using Ceperley-Alder (CA) type parametrization
[d6]For nanocrystalline coatings deposited by dual DC magnetron sputtering, thickness – 0.5–0.7 μm, ~15% amorphous carbon phase; data obtained from nano-indentation measurements
[d7]Calculated on the basis of DFT with GGA using Perdew's functional
[d8]Calculated on the basis of DFT with GGA using the revised version of the Perdew-Burke-Ernzerhof (RPBE) functional
[d9]Calculated on the basis of DFT with GGA using FP-LAPW method and Perdew potential
[e1]Calculated value of G = 196 GPa
[e2]Calculated on the basis of DFT with GGA using the plane-wave pseudopotential code Dacapo and Perdew's functional
[e3]Calculated on the basis of three body force potential (TBFP) model
[e4]Calculated values of G = 179 GPa and κ = 2.96 TPa^{-1}
[e5]From *ab initio* calculations
[e6]Calculated values of G = 120 GPa and κ = 3.86 TPa^{-1}
[e7]Corresponding value G = 171–172 GPa
[e8]Hot-pressed, porosity <2%
[e9]Calculated value of G = 181 GPa
[f1]At ~300 K
[f2]Calculated value of G = 157 GPa
[f3]From first principles calculations
[f4]Calculated value of G = 198 GPa
[f5]Calculated on the basis of DFT with GGA using PW parametrization
[f6]Calculated value of G = 193 GPa
[f7]Calculated value of G = 196 GPa
[f8]Calculated on the basis of DFT with LDA using the Ceperley-Alder-Perdew-Zunger (CAPZ) functional (upper and lower values of E are realized for the external stress applied along one of the crystallographic axes and for the stress along any of the bisector direction in each of the coordinate planes, respectively)

(continued)

Table 4.14 (continued)

[f9]Calculated on the basis of DFT with GGA using PBE functional (upper and lower values of E are realized for the external stress applied along one of the crystallographic axes and for the stress along any of the bisector direction in each of the coordinate planes, respectively)
[g1]Calculated on the basis of Murnaghan equation of state (upper and lower values are given to refer to the results obtained using LDA and GGA, respectively)
[g2]Sintered materials, porosity – 12%, measured at 330 °C
[g3]Calculated on the basis of a modified interaction potential model with covalency ($MIPMC_v$)
[g4]Calculated value of G = 172 GPa
[g5]On the basis of interconsistency of the physical properties of transition metal monocarbides
[g6]Calculated on the basis of DFT using full-potential (linear) augmented-plane-wave plus local orbitals (FP-LAPW + LO) method with PW LDA
[g7]Spark plasma sintered, porosity – 1%
[g8]G = 200 GPa
[g9]By means of extended Hückel tight-binding band electronic structure calculations
[h1]Calculated value of G = 205 GPa
[h2]Calculated on the basis of DFT by projector augmented wave (PAW) method with GGA using PBE functional
[h3]Calculated value of G = 209 GPa
[h4]Sintered, porosity – 4%, non-combined C (graphite) – 1.0%, contents: O $\leq$ 0.05%, N – 0.05%, W – 0.2–0.3%
[h5]Experimentally measured by the method of continuous impression of an indenter
[h6]Calculated on the basis of linearized method of "muffin-tin" orbitals using atomic-sphere approximation (LMTO-ASA)
[h7]Calculated by all-electron pseudopotentials with pseudo-wave functions
[h8]G = 150 GPa
[h9]Calculated on the basis of DFT using PBE functional with GGA by FP-LAPW method
[i1]G = 204 GPa
[i2]Calculated on the basis of DFT using PBE functional with GGA by plane wave pseudo-potential (PPPW) method
[i3]From first principles calculations based on virtual crystal approximation
[i4]G = 192 GPa
[i5]Calculated on the basis of DFT with GGA using PBE functional for zero temperature and pressure conditions
[i6]G = 198 GPa

253, 174 and 251 GPa, respectively. The experimentally measured velocities of ultrasonic waves propagated in niobium monocarbide phases NbC_{1-x} are

for single crystal $NbC_{0.75}$ [369]:

longitudinal velocity (along <001>) V_S, m s^{-1} 8381,
transversal velocity (along <100>) V_T, m s^{-1} 4376,
transversal velocity (along <010>) V_T, m s^{-1} 4377;

for single crystal $NbC_{0.865}$ [369]:

longitudinal velocity (along <110>) V_S, m s^{-1} 8003,
transversal velocity (along <001>) V_T, m s^{-1} 4452,
transversal velocity (along <1$\bar{1}$0>) V_T, m s^{-1} 5394;

for $NbC_{\sim 1.0}$ (summarized on the basis of several sources) [382, 867]

average velocity V_m, m s^{-1} 6450;

and theoretically calculated sound velocities for near-stoichiometric niobium monocarbide $NbC_{\sim 1.0}$ are

by *ab initio* calculations (for zero pressure):

longitudinal velocity V_S, m s^{-1} 8600 [396], 8562 [366, 396, 398], 8698 [375, 396], 8918 (0 K) [402], 7738 (~300 K) [402], 7320 [399], 8440–8570 [409], 8152 [404], 8861 [926];
transversal velocity V_T, m s^{-1} 5119 [396], 4722 [366, 396, 398], 5013 [375, 396], 4960 (0 K) [402], 4261 (~300 K) [402], 3920 [399], 5020–5200 [409], 4615 [404], 5420 [926];
average velocity V_m, m s^{-1} 5662 [396], 5258 [366, 396, 398], 5562 [375, 396], 4370 [399], 5560–5740 [409], 5133 [404], 5975 [926];

by LMTO (linearized method of "muffin-tin" orbitals) calculations [382]:

average velocity V_m, m s^{-1} 6380;

by LMTO-ASA (linearized method of "muffin-tin" orbitals using atomic-sphere approximation) calculations [867]:

average velocity V_m, m s^{-1} 8470;

by TBFP (three body force potential) modeling calculations [399]:

longitudinal velocity V_S, m s^{-1} 8720;
transversal velocity V_T, m s^{-1} 4850;
average velocity V_m, m s^{-1} 5400.

The variations of the elastic properties of niobium monocarbide phases NbC_{1-x} with temperature are shown in Figs. 4.32 and 4.33; for $NbC_{\sim 1.0}$ the average rate of decay of the elastic modulus with temperature $1/E_{298K} \times dE/dT = -1.4 \times 10^{-4}$ K^{-1} [739]; the temperature dependence of Young's modulus E of polycrystalline NbC_{1-x} follows Wachtman's equation, and was expressed by Brenton et al. [193] for materials with $x = 0.03$ in the range from 12 to 2000 °C as

$$E = 460.4 - (5.729 \times 10^{-2})T \exp(-200.5/T), \tag{4.65}$$

while the similar dependence of Coulomb's (shear) modulus G for the same materials and temperature ranges is linear:

$$G = 192.9 - (2.484 \times 10^{-2})T, \tag{4.66}$$

where T is temperature, K; for bulk (compression) modulus K of quasi-stoichiometric niobium monocarbide NbC_{1-x} Ajami and MacCrone [349] proposed the equation similar to Wachtman's for the wide temperature range as well:

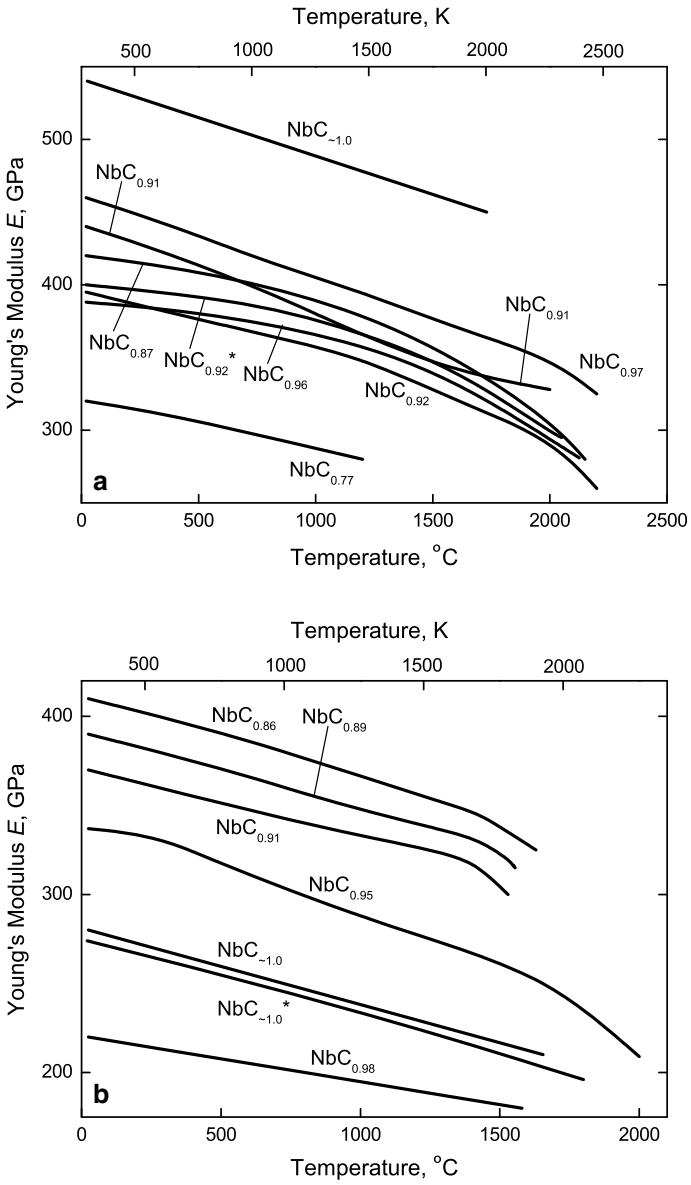

Fig. 4.32 Variations of Young's modulus E with temperature of niobium monocarbide NbC_{1-x} phases: (**a**) – $NbC_{\sim 1.0}$ (recommended for the poreless state by Kotelnikov et al. [5]); $NbC_{0.97}$, $NbC_{0.92}$, $NbC_{0.91}$ and $NbC_{0.77}$ sintered, no porosity correction (measured by Baranov et al. [352]); $NbC_{0.96}$, $NbC_{0.92}$* and $NbC_{0.87}$ sintered, porosity – 7–27%, mean grain size – 10–29 μm, contents: non-combined C – 0.05–0.27%, O – 0.05–0.08%, N – 0.27-0.45%, no porosity correction (measured by Bukatov et al. [28, 361]); (**b**) – $NbC_{\sim 1.0}$, $NbC_{0.98}$, $NbC_{0.91}$, $NbC_{0.89}$ and $NbC_{0.86}$ sintered in vacuum (measured by Avgustinik et al. [354]); $NbC_{0.95}$ (calculated on the basis of hardness-elasticity relationship by Travushkin et al. [270]); $NbC_{\sim 1.0}$* sintered, 18% porosity (measured by Kashtalyan [1054] and recommended by Marmer et al. [170])

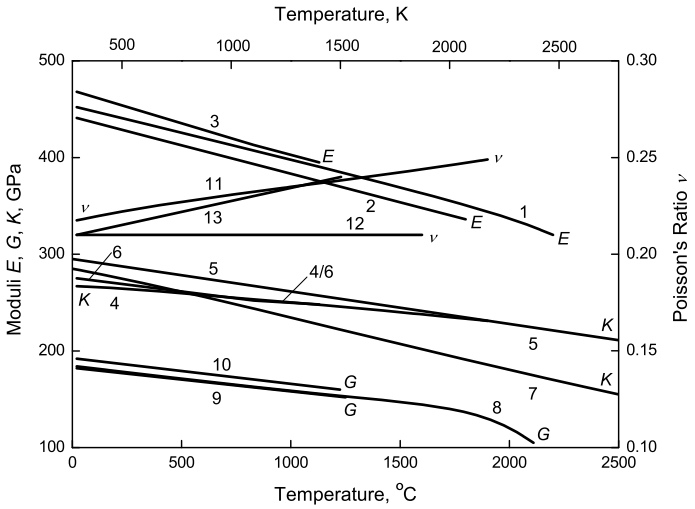

Fig. 4.33 Variations of elastic properties (1–3 – Young's modulus E, 4–7 – bulk (compression) modulus K, 8–10 – Coulomb's (shear) modulus G and 11–13 – Poisson's ratio ν) of near-stoichiometric monocarbide NbC_{1-x} materials with temperature: 1, 4, 8, 11 – hot-pressed $NbC_{0.97}$, 4% porosity (measured by Brenton et al. [193]); 2, 9, 12 – $NbC_{0.97}$, 6% porosity (measured by Speck and Miccioli [7, 357]); 3, 6, 10, 13 – hot-pressed $NbC_{0.95}$, 4% porosity (measured by Jun and Shaffer [355]); 5 – $NbC_{\sim 1.0}$ (first-principles calculated by Cao et al. [356]); 7 – $NbC_{\sim 1.0}$ (first-principles calculated by Hua and Li [955])

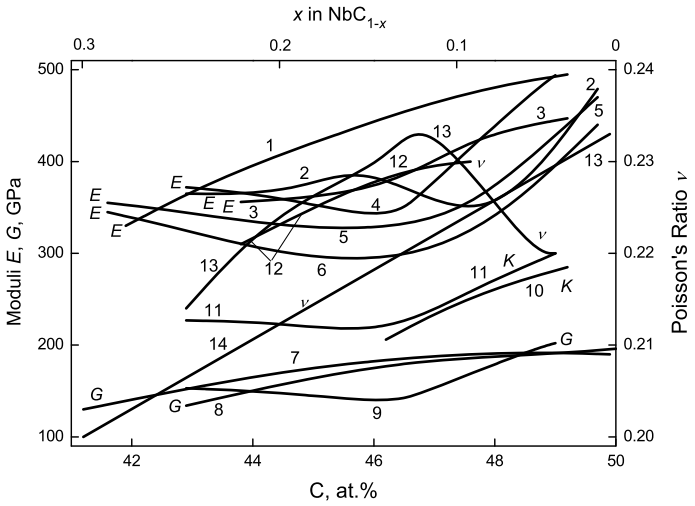

Fig. 4.34 Variations of elastic properties (1–6 – Young's modulus E, 7–9 – Coulomb's (shear) modulus G, 10–11 – bulk (compression) modulus K and 12–14 – Poisson's ratio ν) of niobium monocarbide NbC_{1-x} materials with deviation from stoichiometry within the homogeneity range: 1 – sintered, porosity – 5–10%, corrected to the poreless state [1, 360]; 2 – sintered in vacuum and annealed, porosity – 7–27%, mean grain size – 10–29 μm, contents: non-combined C – 0.05–0.27%, O – 0.05–0.08%, N – 0.27–0.45% [1, 361]; 3, 10, 12 – sintered (in vacuum), E corrected to the poreless state, K and ν calculated [704]; 4, 9, 13 – sintered, porosity – 2.2–12.5%, content O < 0.05% [1, 282]; 5–6 – sintered (5 – 20 °C, 6 – 1000 °C) [2, 28]; 7, 14 – summarized on the basis of several sources [359]; 8 – DFT-calculated [393]; 11 – calculated [883]

$$K = 269 - (1.87 \times 10^{-2})T \exp(-278/T). \tag{4.67}$$

The variations of the elastic properties of niobium monocarbide NbC_{1-x} with deviation from stoichiometry within the homogeneity range are presented in Fig. 4.34. According to Kurlov and Gusev [359], the concentration dependences of Coulomb's (shear) modulus G and Poisson's ratio ν at room temperature have the following forms

$$G = 189.2 + 102.5x - 1001x^2, \tag{4.68}$$

$$\nu = 0.2335 - 0.1169x, \tag{4.69}$$

where x is the value of index in NbC_{1-x} formula. The elastic properties (stiffness coefficients c_{ij} and moduli E, K and Poisson's ratio ν) of various niobium monocarbide NbC_{1-x} phases are listed in Table 4.14; the DFT-calculated stiffness coefficients c_{ij} for α-Nb_2C (*Pnma*) are c_{11} = 424, c_{22} = 431, c_{33} = 410, c_{44} = 74, c_{55} = 137, c_{66} = 117, c_{12} = 128, c_{13} = 141, c_{23} = 113 (in GPa) [154], or c_{11} = 390, c_{22} = 405, c_{33} = 390, c_{44} = 76, c_{55} = 119, c_{66} = 122, c_{12} = 157, c_{13} = 129, c_{23} = 136 (in GPa) [1097] and for β-$Nb_{2+x}C$ (*P(–3)m1*) the values are c_{11} = 426, c_{33} = 526, c_{44} = 79, c_{12} = 129, c_{13} = 108 (in GPa) [405]. The theoretically calculated variations of the elastic moduli E, G and K and Poisson's ratio ν of

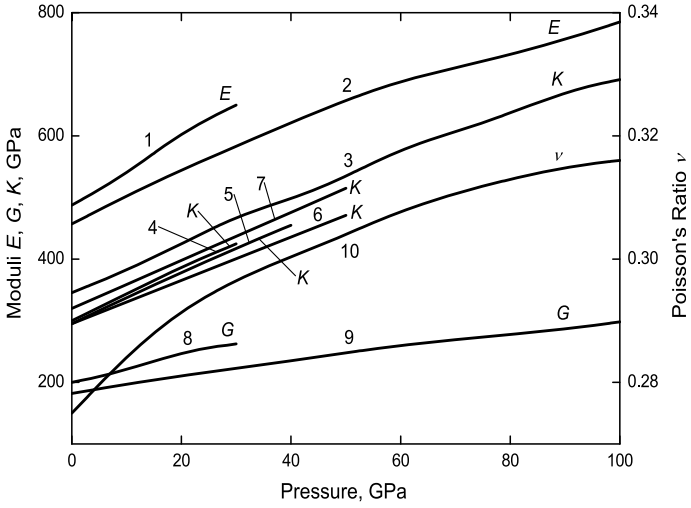

Fig. 4.35 Variations of elastic properties (1–2 – Young's modulus E, 3–7 – bulk (compression) modulus K, 8–9 – Coulomb's (shear) modulus G and 10 – Poisson's ratio ν) of near-stoichiometric niobium monocarbide NbC_{1-x} with higher pressure (computed): 1, 4, 8 – by Soni et al. [396]; 2–3, 9–10 – by Varshney and Shriya [402]; 5 – by Cao et al. [356]; 6 – by Krasnenko and Brik [409] and 7 – by Gautam and Kumar [796]

Table 4.15 Thermal shock/stress resistance (thermal strength) testing data on single phase niobium monocarbide NbC_{1-x} materials

Materials and its characterization	Specimen shape and dimensions, mm	Method of thermal loading (testing)	Experimental (or calculated) results[a]	Reference
$NbC_{\sim 1.0}$, gas-phase deposited, contents: O + N $\leq$ 0.3%	Hollow cylinder: wall thickness – 0.8–1.2, height – 20	Direct resistance heating with 400 K s^{-1}, cooling by inert gas blow, cycle time – 3 min	Disintegration during the first heating-cooling cycle from 2400 to 50 °C	[2, 280]
$NbC_{\sim 1.0}$, sintered, porosity 12%	Hollow cylinder: outer diameter – 48, wall thickness – 0.75	Steady-state heating	R = 43 K[b,c]	[727]
$NbC_{\sim 1.0}$, sintered, porosity – 8%, mean grain size – 15 μm	Rod: diameter – 3, length – 100	Steady-state heating	R = 70 K[b]	[727]
$NbC_{0.93}$, sintered, porosity 10–12%, non-combined C – 0.2%	Ring-shaped disk: outer diameter – 50, inner diameter – 44, height – 10	Steady-state heating along the outer surface	R = 40–50 K (calculated), R' = 1.9 ± 0.3 kJ m^{-1} s^{-1} (experimental)	[2, 3, 728]
$NbC_{0.96}$, sintered, porosity – 4%, contents: O $\leq$ 0.05%, N – 0.05%, W – 0.2–0.3%	Disks with a central hole: inner diameter – 5	Unsteady-state heating along the lateral surface of central hole in molten tin	R = 48 K, R' = 1.2 kJ m^{-1} s^{-1}, R''' = 3.9 MPa^{-1}, R'''' = 0.07 mm	[731]
$NbC_{\sim 1.0}$, sintered, porosity – 4%, non-combined C (graphite) – 1.0%, contents: O $\leq$ 0.05%, N – 0.05%, W – 0.2–0.3%	Disks with a central hole: inner diameter – 5	Unsteady-state heating along the lateral surface of central hole in molten tin	R = 38 K, R' = 1.0 kJ m^{-1} s^{-1}, R''' = 7.6 MPa^{-1}, R'''' = 0.11 mm	[731]
$NbC_{\sim 1.0}$, sintered, porosity – 8%, non-combined C (non-graphitized) – 1.0%, contents: O $\leq$ 0.05%, N – 0.05%, W – 0.2–0.3%	Disks with a central hole: inner diameter – 5	Unsteady-state heating along the lateral surface of central hole in molten tin	R = 61 K, R' = 1.3 kJ m^{-1} s^{-1}, R''' = 4.0 MPa^{-1}, R'''' = 0.13 mm	[731]
$NbC_{\sim 1.0}$, sintered	–	–	Estimated average values: R = 45 K, R' = 1.12 kJ m^{-1} s^{-1}	[735, 736]

(continued)

Table 4.15 (continued)

Materials and its characterization	Specimen shape and dimensions, mm	Method of thermal loading (testing)	Experimental (or calculated) results[a]	Reference
$NbC_{\sim 1.0}$	Linear dimension <30 (area of the beam spot – 100 mm^2)	Thermal pulsed electron beam (energy – 10 keV) with power density – up to 18 kW cm^{-2} and pulse length – up to 1 s	Limiting thermal flux (experimentally determined for thermo-mechanical fracture, recalculated for the initial temperature of 20 °C and pulse length of 0.5 s) – 1.2 kW cm^{-2}	[1055]

[a]Figures-of-merit (criteria) of thermal shock/stress resistance [2, 3, 729, 730, 797]: $R = \sigma_t(1 - \nu)/\alpha E$, $R' = \sigma_t\lambda(1 - \nu)/\alpha E$, $R''' = E/\sigma_t{}^2(1 - \nu)$, $R'''' = \gamma E/\sigma_t{}^2(1 - \nu)$, where σ_t is tensile strength, ν is Poisson's ratio, α is thermal expansion coefficient, E is Young's modulus, λ is thermal conductivity and γ is fracture surface energy

[b]The distribution of R values measured on a representative number of samples (at least 70–90) is characterized by the exponential Weibull relationship; the coefficient of variation agrees closely with similar values obtained from mechanical strength testing

[c]*See also* Table 4.17

near-stoichiometric niobium monocarbide NbC_{1-x} with higher pressure (up to 100 GPa) are shown in Fig. 4.35.

The testing data on thermal shock/stress resistance (thermal strength) of single phase niobium monocarbide NbC_{1-x} materials are given in Table 4.15.

The magnitudes of physico-mechanical (strength, elasticity) properties of near-stoichiometric niobium carbides in the wide range of temperatures are summarized in Addendum in comparison with other ultra-high temperature materials.

4.5 Nuclear Physical Properties

The comprehensive lists of isotopes for niobium Nb and carbon C elements are presented in Tables I-8.4 and I-2.12, respectively. The nuclear physical properties of the elements, including isotopic mass range, total number of isotopes, thermal neutron macroscopic cross sections, moderating ability and capture resonance integral, are given in Table I-A.8.

The thermal neutron macroscopic cross sections Σ_i (*see* Eq. 2.53) of near-stoichiometric niobium carbides $NbC_{1-x}/Nb_{2+x}C$ (for 2200 m s^{-1} neutrons) [5]:

cross section of capture (absorption) Σ_a, cm^{-1}	0.052/0.055,
cross section of scattering Σ_s, cm^{-1}	~0.49/~0.41.

For the estimation of probable damage of niobium monocarbide materials exposed to various types of radiation the parameters of formation and migration of lattice point defects (vacancies and interstitial atoms) are given in Table 4.16. An example of such damage (swelling) produced by fast neutrons ($E > 1$ MeV) in the range of moderate temperatures is shown in Fig. 4.36. The damage produced by fast neutrons fluence up to (2.5–5.0) × 10^{21} cm^{-2} at 130–355 °C results in the

slight failure of near-stoichiometric NbC_{1-x} hot-pressed (or sintered) materials, which is accompanied by ~0.2% relative change of lattice parameter [413]. Lanin [727] has summarized some data on the resistance to irradiation damage of NbC_{1-x} materials in the modeling nuclear rocket engine reactor environment (Table 4.17), which were obtained by testing in reactor IVG-1 [733] and measurements of physical properties after the irradiation. While the geometric volume and shape of the NbC_{1-x} samples are retaining, the irradiation leads to an appreciable growth of the lattice parameter and electrical resistivity accompanied by a minor change of strength and elastic characteristics [314, 727, 734]. According to Dew-Hughes and Jones [799], a neutron ($E \geq 1$ MeV) irradiation dose (1.5×10^{20} cm^{-2}) causes the reduction in superconducting critical temperature T_c of ~6–27%, which is much less than that for the similarly irradiated superconductors of other types. The effect of earlier stages (< 30 s) of nuclear reactor irradiation with neutron flux $\varphi = \sim 1.0 \times 10^{14}$ cm^{-2} s^{-1} and gamma quanta dose of 2.58×10^6 C kg^{-1} on the heat capacity and electrical resistivity of $NbC_{\sim 1.0}$ materials, when the gamma quanta influence dominates, was studied by Taubin [800]. Gusev et al. [945] conclude that at other equal conditions NbC_{1-x} phases with a larger deficit on

Table 4.16 Parameters of formation and migration of lattice point defects (vacancies and interstitial atoms) in near-stoichiometric niobium monocarbide[a]

Defect	Metal sublattice				Non-metal sublattice			
	E_f, eV	E_m, eV	S_f/k_B	S_m/k_B	E_f, eV	E_m, eV	S_f/k_B	S_m/k_B
Vacancy[b] [411, 944]	4.09[c]	3.41	4.89	4.08	1.36	5.44	1.63	6.52
Vacancy[d] [706, 944]	3.64[c]	3.19	4.37	3.82	1.06	2.98	1.27	3.57
Vacancy[e] [393]	4.10[c]	–	–	–	−0.42[f]	3.60[g]	–	–
Vacancy[h] [938]	1.94	–	–	–	–	–	–	–
Vacancy[i] [3]	1.98	–	–	–	–	–	–	–
Interstitial atom[b] [411]	19.72	0.68	23.63	0.81	6.12	0.68	7.33	0.81

[a]Denoted: E_f – defect formation energy, E_m – defect migration energy, S_f – defect formation entropy, S_m – defect migration entropy, k_B – Boltzmann constant

[b]Calculated on the basis of bonding model (relaxation displacement of atoms surrounding a defect is not taking into account)

[c]According to Ordanyan et al. [76] the values of vacancy formation energy decreases with increasing defectiveness of the non-metal sublattice from 4.8 eV for $NbC_{\sim 1.0}$ to 3.6 eV for $NbC_{0.72}$, Storms et al. [147] determined 2.62 eV for $NbC_{0.75}$ and Yu and Davis [630] – 3.30 eV for $NbC_{0.77}$

[d]Calculated on the basis of elastic continuum model

[e]*Ab initio* simulation computed from the largest super-cell on the basis of the PAW pseudopotential using the PBE formulation

[f]The computed values of vacancy formation energy on the basis of the *ab initio* pseudopotential approach for $NbC_{0.875}$ and $NbC_{0.75}$ are −0.13 and −0.51 eV, respectively [678]

[g]The calculated value of activation energy for carbon self-diffusion in niobium monocarbide $E_A = 350$ kJ mol^{-1} (*see also* Sect. 4.6, Table 4.21)

[h]Calculated on the basis of data on the yield strength determined through hardness and elasticity measured experimentally

[i]Experimentally determined

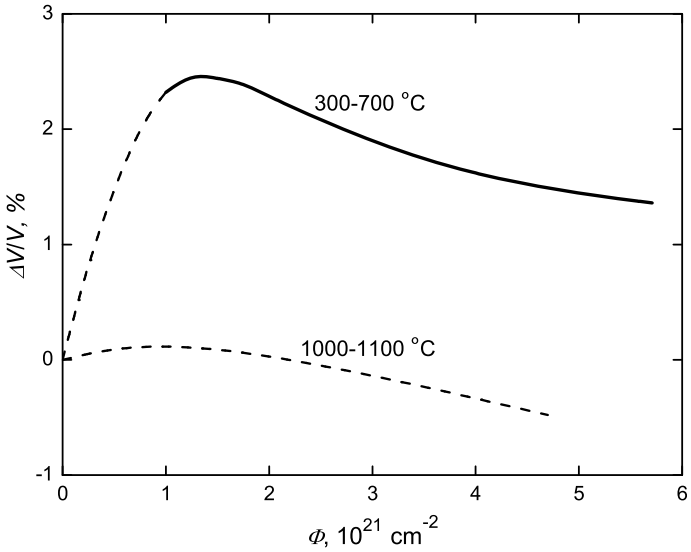

Fig. 4.36 The change in macroscopic volume (swelling/shrinkage) with fast neutron (>1 MeV) fluence Φ for produced by ceramic technologies (hot-pressed, slip-cast, explosion-pressed) near-stoichiometric niobium monocarbide NbC_{1-x} materials irradiated at different temperatures [411, 413, 798, 1056, 1057]

carbon are more radiation resistant. Deb et al. [1059] used 59.5 keV gamma radiation from ^{241}Am source for the measurement of Compton profile of niobium monocarbide NbC_{1-x}.

At room temperature the damage produced on quasi-stoichiometric monocarbide reactively sputtered films (thickness $\sim$0.2 μm) by 600 keV Ar ions and 200 keV He ions irradiations in the fluence range of 0.01–100 dpa (or 10^{13}–10^{17} cm^{-2} and 10^{15}–10^{19} cm^{-2} for Ar and He ions, respectively) leads to the lattice expansion at the first irradiation state (maximum value of expansion 0.4% and maximum static atomic displacement of metal atoms 8 pm are corresponding to the fluence of $\sim$0.4 dpa), which is mainly attributed to interstitial metal defects including antisite defects, and subsequent contraction at the second irradiation state (minimum value of -0.5% corresponding to $\sim$40 dpa), which is attributed to the clustering or precipitation of the interstitial metal defects; 200 keV He irradiated $NbC_{0.98}$ single crystal showed the similar behaviour and was characterized at the first irradiation state by small displacements in the metal sublattice and considerable disorder of the non-metal sublattice [414–416]. The superconducting transition temperature T_c of quasi-stoichiometric monocarbide thin films decreases continuously from above 11 K to around 4 K and saturates in the fluence range of 10–100 dpa. The effects of both 600 keV Ar ions and 200 keV He ions irradiations on the changes of the lattice parameter and T_c depressions are nearly the same at 0.01–100 dpa. During

Table 4.17 Structural and physico-mechanical properties of near-stoichiometric niobium monocarbide NbC_{1-x} materials before and after irradiation exposure at different temperatures [2, 727]

Characteristics	Unit	Measured values		
		Initial samples	Irradiated samples[a]	
		20 °C	150 °C	1100 °C
Density	g cm^{-3}	7.40	7.40	7.40
Lattice parameter	nm	0.4471	0.4488	0.4472
Electrical resistivity	μΩ m	0.50	0.90	0.50
Flexural strength:				
average	MPa	340	340	380
minimal	MPa	290	300	330
maximal	MPa	400	380	400
Young's modulus	GPa	480	465	490
Microhardness	GPa	13.7	24.5	15.7
load[b]	N	1.18	0.98	–
Thermal stress resistance criterium R[c]	K	70	130	135

[a]The neutron flux – 10^{12}–10^{15} cm^{-2} s^{-1}
[b]The minimum load during the microhardness tests, at which all the indentations were cracking
[c]*See* notes to Table 4.15

the annealing process of $NbC_{\sim 1.0}$ films irradiated at the first state, stacking fault formation is inferred around 250 °C and recovery of the displacements of the metal atoms has been observed around 700 °C [237, 414]. Employing a positron annihilation method for the studies of atomic defects in the irradiated refractory carbides at low temperatures, Rempel et al. [945, 1058] proved that the recombination of interstitial niobium atoms and vacancies in $NbC_{0.84}$ and $NbC_{0.98}$ occurred at temperatures below 85 K.

Nuclear physical properties of niobium carbides in comparison with other ultra-high temperature materials are also given in Addendum.

4.6 Chemical Properties and Materials Design

The comprehensive data on the chemical properties, compatibility (in the connection with both environmental resistance and composite materials design) and interaction behaviour of niobium mono- and semicarbide phases at elevated, high and ultra-high temperatures with elements (metals, non-metals) are summarized in Table 4.18, with refractory compounds – in Table 4.19 and with gaseous media – in Table 4.20. The data on the oxidation resistance of niobium monocarbide materials listed there are also accompanied by the graphic information in Figs. 4.37 and 4.38; the isothermal oxidation kinetics of near-stoichiometric NbC_{1-x} materials given there can be considered in the context of the ridge-effect model proposed by

Table 4.18 Chemical interaction and/or compatibility of niobium carbide phases with elements (metals, non-metals) at elevated, high and ultra-high temperatures, including solid matters and molten media (reaction systems are given mainly in alphabetical order)[a]

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	Reference
NbC_{1-x}–Ag–Cu			*See* Table 4.21	
NbC_{1-x}–Al	–	700–800	No chemical interaction	[417–422, 561, 569, 589, 900, 901, 912, 914, 915, 919, 948, 984, 992, 1016, 1081–1083]
	–	900	Weak interaction	
	–	–	Formation of Nb_2AlC, α/β-Nb_4AlC_{3-x} and Nb_5AlC_4 ($M_{n+1}AX_n$-phases) and Nb_3Al_2C	
			See also section NbC_{1-x}–Al_4C_3 in Table 4.19	
			See also Table 4.21	
			See also section C–Al–Nb in Table I-2.14	
NbC_{1-x}–Al–Co	Ar	1350	Formation of η-$Nb_3(Co,Al)_3C_{1\pm x}$ phases	[1084]
NbC_{1-x}–Al–Cr	Ar	1350	Formation of η-$Nb_3(Cr,Al)_3C_{1\pm x}$ phases	[1084]
NbC_{1-x}–Al–Cu	Ar	1350	Formation of η-$Nb_3(Cu,Al)_3C_{1\pm x}$ phases	[1084]
NbC_{1-x}–Al–Fe	–	1000	Particles of NbC_{1-x} are rather stable in the Fe – 15–26 at.% Al alloys during long-term exposures, although the formation of ε-$(Fe,Al)_{2\pm x}Nb$ was observed; the solubility of Fe in NbC_{1-x} varies from 0.9 to 5.0 at.% and that of Al is very low	[1011, 1025, 1026, 1084]
	Ar	1350	Formation of η-$Nb_3(Fe,Al)_3C_{1\pm x}$ phases	
	Vacuum, 4 Pa	1420	The liquid-phase sintering of powdered $NbC_{\sim 1.0}$ (mean particle size – 1.2 μm) with 12% addition of Fe – 25 at.% Al alloy (mean particle size <40 μm) as a metallic binder leads to the formation of hard alloy system	
NbC_{1-x}–Al–Mg	–	800–1000	Most of NbC_{1-x} particles decompose in molten Al – 1 mas.% Mg alloy	[1004]
NbC_{1-x}–Al–Mn	Ar	1350	Formation of η-$Nb_3(Mn,Al)_3C_{1\pm x}$ phases	[875]
NbC_{1-x}–Al–Ni	Ar	1350	Formation of η-$Nb_3(Ni,Al)_3C_{1\pm x}$ phases	[875]
NbC_{1-x}–Al–V	Ar	1350	Formation of η-$Nb_3(V,Al)_3C_{1\pm x}$ phases	[875]
NbC_{1-x}–Al–Zn	Ar	1350	Formation of η-$Nb_3(Zn,Al)_3C_{1\pm x}$ phases	[875]
β-$Nb_{2+x}C$–Al	–	700–800	No chemical interaction	[417–419, 569]
	–	900	Weak interaction	
			See also section C–Al–Nb in Table I-2.14	
NbC_{1-x}–SiC–TiC_{1-x}–Al	–	–	Formation of solid solutions $(Ti_{1-x}Nb_x)_3(Si_{1-x}Al_x)C_2$ based on Ti_3SiC_2	[903]
			See also section NbC_{1-x}–Al_4C_3–SiC–TiC_{1-x} in Table 4.19	

(continued)

Table 4.18 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	Reference
NbC_{1-x}–TiC_{1-x}–Al	–	–	Formation of $(Nb,Ti)_2AlC$, $(Nb,Ti)_4AlC_3$ and $(Nb,Ti)_5AlC_4$ ($M_{n+1}AX_n$-phase solid solutions); $(Nb_{0.5}Ti_{0.5})_2AlC$, $(Nb_{0.5}Ti_{0.5})_5AlC_4$, $(Nb_{1-x}Ti_x)_3AlC$ and $(Nb_{1-x}Ti_x)_4AlC_3$ ($0 \le x \le 0.3$) were synthesized	[421, 422, 561, 573, 912–914, 954, 984, 1016, 1061]
			See also section NbC_{1-x}–Al_4C_3–TiC_{1-x} in Table 4.19	
			See also section C–Al–Nb–Ti in Table I-2.14	
NbC_{1-x}–VC_{1-x}–Al	–	–	Formation of $(Nb,V)_2AlC$ and $(Nb,V)_4AlC_3$ ($M_{n+1}AX_n$-phase solid solutions); compositions $(Nb_{0.5}V_{0.5})_2AlC$ and $(Nb_{0.5}V_{0.5})_4AlC_3$ were synthesized	[421, 422, 573, 912, 1016]
			See also section C–Al–Nb–V in Table I-2.14	
NbC_{1-x}–ZrC_{1-x}–Al	–	–	Formation of $(Nb,Zr)_2AlC$ ($M_{n+1}AX_n$-phase solid solution); $(Nb_{0.6}Zr_{0.4})_2AlC$ and $(Nb_{0.8}Zr_{0.2})_2AlC$ were synthesized	[419, 421, 422, 573, 912, 1014–1016]
			See also section C–Al–Nb–Zr in Table I-2.14	
NbC_{1-x}–As	–	–	Formation of Nb_2AsC ($M_{n+1}AX_n$-phase)	[421, 422]
			See also section C–As–Nb in Table I-2.14	
NbC_{1-x}–Au			*See* Table 4.21	
NbC_{1-x}–β-B	–	>2250	Formation of $NbB_{2\pm x}$	[1, 5, 473, 474]
			See also section C–B–Nb in Table I-2.14	
β-$Nb_{2+x}C$–β-B			*See* section C–B–Nb in Table I-2.14	
NbC_{1-x}–Be	Vacuum	1500	Formation of the terminal monocarbide (cubic) solid solutions of substitutional-interstitial type	[985–987]
NbC_{1-x}–Bi			*See* Table 4.21	
NbC_{1-x}–C	Vacuum	<3150	NbC_{1-x} is compatible in contact with C (graphite) parts and articles	[1, 5, 58, 146, 335, 566, 567, 810, 874, 916, 922, 1077]
	–	~3250–3300	Eutectic NbC_{1-x}–C (graphite)	
			See also section C–Nb in Table I-2.13	
NbC_{1-x}–C–Fe			*See* Table 4.21	
NbC_{1-x}–C–Fe–Mn			*See* Table 4.21	

(continued)

Table 4.18 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	Reference
NbC_{1-x}–α/ε-Co	–	1100	Formation of η-Nb_4Co_2C, or η-Nb_3Co_3C	[1, 5, 69, 81, 423, 554, 999, 1011, 1012, 1076, 1085]
	–	1250	The max. solid solubility of NbC_{1-x} in α-Co is ∼3–6 mol.% (?)	
	–	1350–1400	Metallic Co on NbC_{1-x} plate instantaneously penetrates into the carbide and forms a eutectic (the process is identical in the case of both sintered and fused carbide materials)	
	–	∼1360–1380	Eutectic NbC_{1-x}–α-Co; the max. solubility of NbC_{1-x} in α-Co is ∼1.5 mol.% and that of Co in NbC_{1-x} is 8 mol.%	
	Vacuum, 4 Pa	1420	The liquid-phase sintering (1 h exposure) of powdered $NbC_{\sim 1.0}$ (mean particle size – 1.2 μm) – 20 mol.% Co (mean particle size – 0.6 μm) composition leads to the formation of hard alloy based on $(Nb_{0.99\pm0.01}Co_{0.01\pm0.005})C_{0.89}$ phase with $(Co_{0.86}Nb_{0.03\pm0.03}C_{0.11})$ binder	
	–	1400–1500	Intensive interaction (in the contact zone carbide has a changed lattice parameter)	
	–	∼1400	The max. solubility of NbC_{1-x} in molten Ni is 5.0 mol.%	
	–	1450	The solubility of Nb in Co-based binder phase in equilibrium with C is 3.5–4.0%	
			See also Table 4.21	
			See also section C–Co–Nb in Table I-2.14	
NbC_{1-x}–Co–Si			*See* Table 4.21	
NbC_{1-x}–HfC_{1-x}–Co			*See* section HfC_{1-x}–NbC_{1-x}–Co in Table 3.19	
			See also section C–Co–Hf–Nb in Table I-2.14	
NbC_{1-x}–β-$Mo_{2\pm x}C$–Co	Vacuum, 4 Pa	1420	The liquid-phase sintering (1 h exposure) of powdered $NbC_{\sim 1.0}$ (1.2 μm) – 2.5 mol.% β-$Mo_{2\pm x}C$ (1.2 μm) – 20 mol.% Co (0.6 μm) composition (mean particle sizes are given in brackets) leads to the formation of a hard alloy with core-rim microstructures and approximate compositions of hard phase – $(Nb_{0.95\pm0.04}Mo_{0.04}Co_{0.01})C_{1.06\pm0.06}$ and metallic binder phase – $(Co_{0.81\pm0.03}Mo_{0.05}Nb_{0.01}C_{0.13\pm0.03})$	[1011]
NbC_{1-x}–β-$Mo_{2\pm x}C$–TiC_{1-x}–δ-$TiN_{1\pm x}$–δ-$WC_{1\pm x}$–Co–Ni	Ar	1360	Formation of $(Nb,Ti,W,Mo)(C,N)_{1-x}$–$Ti(C,N)_{1-x}$–(Co,Ni,Mo,Nb) quasi-three-phase system with core-rim microstructures; the solubility of Nb in $(Nb,Ti,W,Mo)(C,N)_{1-x}$ is limited, so the redundant amounts of Nb dissolves in the metallic binder	[1072]
NbC_{1-x}–TaC_{1-x}–Co			*See* section TaC_{1-x}–NbC_{1-x}–Co in Table 2.21	
			See also section C–Co–Nb–Ta in Table I-2.14	

(continued)

Table 4.18 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	Reference
NbC_{1-x}–TaC_{1-x}–TiC_{1-x}–δ-$WC_{1\pm x}$–Co			*See* section TaC_{1-x}–NbC_{1-x}–TiC_{1-x}–δ-$WC_{1\pm x}$–Co in Table 2.21	
			See also section C–Co–Nb–Ta–Ti–W in Table I-2.14	
NbC_{1-x}–TaC_{1-x}–TiC_{1-x}–δ-$TiN_{1\pm x}$–Co			*See* section TaC_{1-x}–NbC_{1-x}–TiC_{1-x}–δ-$TiN_{1\pm x}$–Co in Table 2.21	
NbC_{1-x}–TaC_{1-x}–TiC_{1-x}–δ-$TiN_{1\pm x}$–Co–Ni			*See* section TaC_{1-x}–NbC_{1-x}–TiC_{1-x}–δ-$TiN_{1\pm x}$–Co–Ni in Table 2.21	
NbC_{1-x}–TaC_{1-x}–TiC_{1-x}–δ-$TiN_{1\pm x}$–δ-$WC_{1\pm x}$–Co			*See* section TaC_{1-x}–NbC_{1-x}–TiC_{1-x}–δ-$TiN_{1\pm x}$–δ-$WC_{1\pm x}$–Co in Table 2.21	
NbC_{1-x}–TaC_{1-x}–TiC_{1-x}–δ-$TiN_{1\pm x}$–VC_{1-x}–δ-$WC_{1\pm x}$–Co–Ni–Mo			*See* section TaC_{1-x}–NbC_{1-x}–TiC_{1-x}–δ-$TiN_{1\pm x}$–VC_{1-x}–δ-$WC_{1\pm x}$–Co–Ni–Mo in Table 2.21	
NbC_{1-x}–TiC_{1-x}–Co	–	~1390–1410	Eutectic $(Nb,Ti)C_{1-x}$ (x = 0.10–0.14)–α-Co	[423, 803, 1003]
			See also section C–Co–Nb–Ti in Table I-2.14	
NbC_{1-x}–VC_{1-x}–Co	–	~1310–1350	Eutectic $(Nb,V)C_{1-x}$ (x = 0.18)–$(V,Nb)C_{1-x}$ (x = 0.12)–α-Co	[10, 424, 425, 1011]
	Vacuum, 4 Pa	1420	Liquid-phase sintering (1 h exposure) of powdered $NbC_{\sim 1.0}$ (1.2 μm) – 7.5 mol.% VC_{1-x} (1.2 μm) – 19 mol.% Co (0.6 μm) composition (mean particle sizes are given in brackets) leads to the formation of hard alloy based on $(Nb_{0.92\pm0.02}V_{0.07\pm0.02}Co_{0.01})C_{0.92}$ phase with $(Co_{0.80\pm0.01}V_{0.04}Nb_{0.02\pm0.01}C_{0.14\pm0.01})$ binder	
			See also section C–Co–Nb–V in Table I-2.14	
NbC_{1-x}–VC_{1-x}–δ-$WC_{1\pm x}$–Co	–	1400	The formation of $Co_{3+x}W_{3-x}C$ is suppressed by the presence of NbC_{1-x} and VC_{1-x}	[977]
NbC_{1-x}–δ-$WC_{1\pm x}$–Co	–	1200–1450	General considerations of the $(Nb,W)C_{1-x}$–δ-$(W,Nb)C_{1\pm x}$–α-Co system are presented	[426, 427, 838, 952, 977, 978, 1011]
	–	1400	The solid solubility of Nb in η_2-$(Co,W)_6C$ is corresponding to $Co_{3.04}W_{2.81}Nb_{0.16}C$ composition	
	Vacuum, 4 Pa	1420	Liquid-phase sintering (1 h exposure) of powdered $NbC_{\sim 1.0}$ (1.2 μm) – 2.5 mol.% δ-$WC_{1\pm x}$ (50 nm) – 20 mol.% Co (0.6 μm) composition (mean particle sizes are given in brackets) leads to the formation of hard alloy based on $(Nb_{0.96}W_{0.02}Co_{0.02})C_{1.07}$ phase with $(Co_{0.79\pm0.03}W_{0.03}C_{0.18\pm0.03})$ binder	
			See also section C–Co–Nb–W in Table I-2.14	

(continued)

Table 4.18 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	Reference
NbC_{1-x}–ZrC_{1-x}–Co	–	~1380–1410	Eutectic $(Nb,Zr)C_{1-x}$ (x = 0.18–0.19)–α-Co	[423]
			See also section C–Co–Nb–Zr in Table I-2.14	
β-$Nb_{2+x}C$–Co			*See* section C–Co–Nb in Table I-2.14	
NbC_{1-x}–Cr	–	1050	The max. solubility of Cr in NbC_{1-x} is <5 at.%; NbC_{1-x} is in equilibrium with all Cr carbides and intermetallide and metallic Cr	[52, 69, 429–432, 465, 553, 932]
		1640	Eutectic NbC_{1-x}–Cr; the max. solubility of Cr in NbC_{1-x} is ~3–9 mol.% and that of NbC_{1-x} in Cr is ~2 mol.%	
	–	≥1640	Eutectic $(Nb,Cr)C_{1-x}$ ($x \approx 0.1$–0.2)–(Cr_yNb_z) (y = 0.98, z = 0.02)	
			See also section C–Cr–Nb in Table I-2.14	
NbC_{1-x}–Cr–Fe–Ni	Vacuum, 4 Pa	1420	The liquid-phase sintering of powdered $NbC_{\sim 1.0}$ (mean particle size–1.2 μm) with 12% addition of Fe – 16–18% Cr – 10–14% Ni – 2–3% Mo stainless steel (mean particle size – 22 μm) as a metallic binder leads to the formation of hard alloy (cermet) system	[1011]
α-Nb_2C–Cr	–	1050	The max. solubility of Cr in β-$Nb_{2+x}C$ is 6 at.% (?); α-Nb_2C phase is in equilibrium with λ_2-$NbCr_{2\pm x}$, NbC_{1-x} and metallic Nb and Cr	[52, 553, 585, 932]
			See also section C–Cr–Nb in Table I-2.14	
β-$Nb_{2+x}C$–Cr	–	1660–1690	The max. solubility of Cr in β-$Nb_{2+x}C$ is ~2 at.%; β-$Nb_{2+x}C$ phase is in equilibrium with λ_1/λ_2-$NbCr_{2\pm x}$, NbC_{1-x} and metallic Nb	[52, 428, 431, 553, 585, 932]
			See also section C–Cr–Nb in Table I-2.14	
NbC_{1-x}–β-$Mo_{2\pm x}C$–TiC_{1-x}–VC_{1-x}–Cr–Ni	Vacuum	1470–1550	The formation of two-phase $(Ti,Nb,V,Mo,Cr)C_{1-x}$–(Ni,Cr,Mo,Ti) hard alloys (cermets)	[1074]
NbC_{1-x}–Cs	Cs vapour, ~13 Pa	~2000	No effect upon structural characteristics of NbC_{1-x} and its thermionic emission parameters (some tens of hours exposure)	[260, 262, 555]
	Cs vapour	–	Chemical reaction with deposited by vacuum evaporation carbide layers (composition and structure uncontrolled) is reported	
NbC_{1-x}–Cu	–	1100–1300	No chemical interaction	[1, 447]
			See also Table 4.21	
NbC_{1-x}–Cu–Nb			*See* Table 4.21	
NbC_{1-x}–Cu–Nb–W			*See* Table 4.21	

(continued)

Table 4.18 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	Reference
NbC_{1-x}–α/γ/δ-Fe	–	~1150	Eutectic NbC_{1-x}–γ-Fe–C	[1, 5, 67, 69, 81, 436–439, 442, 554, 819, 836, 956, 979, 1029, 1078]
	–	1250	The max. solid solubility of NbC_{1-x} in γ-Fe is ~1–2 mol.%	
	–	≤ ~1370	NbC_{1-x} is in equilibrium with ε-$Fe_{2\pm x}Nb$ and γ/δ-Fe	
	–	1390–1440	Metallic Fe on NbC_{1-x} plate instantaneously penetrates into the carbide and forms a eutectic (the process is identical in the case of both sintered and fused carbide materials)	
	–	~1400–1450	Eutectic NbC_{1-x}–δ-Fe	
	–	1450–1550	Intensive interaction (in the contact zone carbide has a changed lattice parameter)	
	–	–	DFT calculated semicoherent interfacial energy at relaxed interface – 0.32 J m^{-2} *See also* Table 4.21 *See also* section C–Fe–Nb in Table I-2.14	
NbC_{1-x}–Fe–Mn	–	–	General consideration of the system *See also* Table 4.21	[980–982]
NbC_{1-x}–Fe–Ni	–	–	General consideration of the system *See also* Table 4.21	[802]
NbC_{1-x}–δ-NbN_{1-x}–TiC_{1-x}–δ-$TiN_{1\pm x}$–Fe	–	1000–1200	Miscibility gap in the NbC_{1-x}–TiC_{1-x}–δ-NbN_{1-x}–δ-$TiN_{1\pm x}$ system in the presence of Fe is determined *See also* section C–Fe–N–Nb–Ti in Table I-2.14	[494, 817, 899]
NbC_{1-x}–δ-NbN_{1-x}–TiC_{1-x}–δ-$TiN_{1\pm x}$–Fe–Ni	–	–	$(Nb_{0.23}Ti_{0.77})(C_{0.42–0.50}N_{0.48–0.50})$ – 25–55% Fe – 30–60% Ni – 2–17% Cr – 2–17% Mo hard metals (cermets) were designed on the basis of wettability/interaction measurements	[1062–1066]
NbC_{1-x}–TiC_{1-x}–δ-$TiN_{1\pm x}$–Fe	–	1200	General consideration of the $(Nb,Ti)(C,N)_{1-x}$–γ-Fe system *See also* section C–Fe–N–Nb–Ti in Table I-2.14	[434]
NbC_{1-x}–TiC_{1-x}–VC_{1-x}–Fe	–	1000–1200	Miscibility gap in the NbC_{1-x}–TiC_{1-x}–VC_{1-x} system in the presence of Fe is determined	[494, 817, 899]
NbC_{1-x}–VC_{1-x}–Fe	–	~1000–1600	Miscibility gap in the NbC_{1-x}–VC_{1-x} system in the presence of Fe is determined	[435, 493, 494, 801, 899]
		1350–1450	Eutectic $(Nb,V)C_{1-x}$–δ-Fe *See also* section C–Fe–Nb–V in Table I-2.14	
NbC_{1-x}–δ-NbN_{1-x}–VC_{1-x}–δ-VN_{1-x}–Fe	–	1000–1200	Miscibility gap in the NbC_{1-x}–VC_{1-x}–δ-NbN_{1-x}–δ-VN_{1-x} system in the presence of γ-Fe is determined; monocarbonitride phases $(Nb_{0.12–0.88}V_{0.14–0.91})(C_{0.43–0.97}N_{0.03–0.57})$ were revealed experimentally	[493, 494, 899, 983]

(continued)

Table 4.18 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	Reference
NbC_{1-x}–δ-NbN_{1-x}–δ-$TiN_{1\pm x}$–VC_{1-x}–δ-VN_{1-x}–Fe	–	1000	General consideration of the $(Nb,Ti,V)(C,N)_{1-x}$–$(V,Nb,Ti)(C,N)_{1-x}$–γ-Fe system *See also* section C–Fe–N–Nb–Ti–V in Table I-2.14	[435]
NbC_{1-x}–δ-NbN_{1-x}–Fe	–	–	General consideration of the $Nb(C,N)_{1-x}$–γ/δ-Fe system *See also* section C–Fe–N–Nb in Table I-2.14	[433]
NbC_{1-x}–δ-NbN_{1-x}–δ-$TiN_{1\pm x}$–γ-Fe	–	1200	General consideration of the $(Nb,Ti)(C,N)_{1-x}$–γ-Fe system *See also* section C–Fe–N–Nb–Ti in Table I-2.14	[434]
β-$Nb_{2+x}C$–$\alpha/\gamma/\delta$-Fe			*See* section C–Fe–Nb in Table I-2.14	
NbC_{1-x}–Fe_3Al	Vacuum	1420	The components are compatible to each other as cermet/hardmetal constituents	[1011, 1012]
NbC_{1-x}–Ga	–	–	Formation of Nb_2GaC ($M_{n+1}AX_n$-phase), or Nb_5GaC_x ($x < 1$, ?) *See also* Table 4.21 *See also* section C–Ga–Nb in Table I-2.14	[81, 421, 422, 564, 747]
NbC_{1-x}–Ge			*See* Table 4.21	
NbC_{1-x}–Hf			*See also* section C–Hf–Nb in Table I-2.14	
β-$Nb_{2+x}C$–α/β-Hf	–	1200–2050	The max. solid solubility of Hf in β-$Nb_{2+x}C$ is ~5 at.% *See also* section C–Hf–Nb in Table I-2.14	[5, 69, 476-478]
NbC_{1-x}–In	–	–	Formation of Nb_2InC ($M_{n+1}AX_n$-phase) *See also* Table 4.21 *See also* section C–In–Nb in Table I-2.14	[421, 422, 456]
NbC_{1-x}–Ir	–	1500	NbC_{1-x} is in equilibrium with $Ir_{3\pm x}Nb$ and (Nb,Ir) alloy *See also* section C–Ir–Nb in Table I-2.14	[69]
β-$Nb_{2+x}C$–Ir			*See* section C–Ir–Nb in Table I-2.14	
NbC_{1-x}–K	K vapour	~2000	No chemical reaction observed, no effect upon structural characteristics of NbC_{1-x} and its thermionic emission parameters (some tens of hours exposure)	[262, 555, 806]
NbC_{1-x}–$\alpha/\beta/\gamma/\delta$-Mn	–	700–1000	NbC_{1-x} is in equilibrium with metallic Mn, intermetallide $Mn_{2\pm x}Nb$ and all Mn carbides	[1, 585]
	–	1350	Formation of eutectic alloys and solid solutions (in the contact zone carbide has a changed lattice parameter) *See also* Table 4.21 *See also* section C–Mn–Nb in Table I-2.14	

(continued)

Table 4.18 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	Reference
NbC_{1-x}–Mo	–	1600	Under pressure 5 MPa the diffusion welded joint between NbC_{1-x} and Mo parts is produced (5–15 min exposure)	[1, 5, 57, 69, 82, 170, 441, 442, 464, 465, 565, 697]
	Vacuum	1700–1800	The initiation of reaction between powdered NbC_{1-x} and compact metallic Mo (exposure 5 h) with the formation of Mo-C solid solution, β-$Mo_{2\pm x}C$ and β-$Nb_{2+x}C$	
	Vacuum, 0.01 Pa	1800	The initiation of reaction between sintered dense NbC_{1-x} ($x \approx 0$) and compact metallic Mo (exposure 2 h)	
	–	1900	The max. solid solubility of Mo is ~45 at.%	
	Vacuum	2000–2200	Intensive contact interaction (exposure >2 h) between powdered NbC_{1-x} and compact metallic Mo *See also* section C–Mo–Nb in Table I-2.14	
β-$Nb_{2+x}C$–Mo	–	1900	The max. solid solubility of Mo is ~3 at.% *See also* section C–Mo–Nb in Table I-2.14	[54, 57, 441]
NbC_{1-x}–TiC_{1-x}–Mo–Ni	Vacuum, 0.05 Pa	1300–1450	The interaction of $Nb_{0.24}Ti_{0.76}C_{0.96}$ phase with Ni – 20% Mo melt is characterized by its dissolution and subsequent precipitation of $Nb_{\sim 0.30}Ti_{0.45-0.50}Mo_{0.20-0.25}C_{1-x}$ (from the melt saturated with Nb, Ti and C)	[815, 989, 1075, 1088, 1089]
NbC_{1-x}–TiC_{1-x}–δ-$TiN_{1\pm x}$–Mo–Ni	Vacuum, 0.01 Pa	1450	The interaction of complex carbonitrides $Nb_{0.03-0.40}Ti_{0.60-0.97}C_{0.35-0.60}N_{0.40-0.65}$ with molten Ni – 20% Mo alloys leads to the accelerated incongruent dissolution of the components, preferentially – C and Nb; after the cooling metallic melts contain primary precipitates of $(Nb_{0.73}Mo_{0.16}Ti_{0.11})(C,N)_{1-x}$ and intermetallide $Ni_{3\pm x}Nb$ phases	[1068]
NbC_{1-x}–Mo–Re	Vacuum	2450	No interaction in the contact zone (1 h exposure) between Mo – 40 mas.% Re alloy and dense carbide materials	[451]
	Vacuum	2600	Interaction in the contact zone (10 min. exposure) between the compact dense materials	
	Vacuum or He	>2625	Being absorbed into the pores molten alloy dissolves (>25 vol.%) in NbC_{1-x} (17% porosity) materials completely *See also* sections C–Mo–Nb and C–Nb–Re in Table I-2.14	
NbC_{1-x}–Na	Na vapour	–	Adsorption of Na on the single crystal surface NbC_{1-x} (111) was studied	[807]
NbC_{1-x}–Nb	–	1300	Under pressure 5 MPa the diffusion welded joint between NbC_{1-x} and Nb parts is produced (5–15 min exposure).	[1, 5, 6, 67, 82, 170, 458, 463, 465, 929]
	Vacuum	1300–2200	The interaction of powdered carbide with metal results in the formation of Nb–C solid solutions and β-$Nb_{2+x}C$ (5 h exposure) *See also* section C–Nb in Table I-2.13	
β-$Nb_{2+x}C$–Nb		~2340–2360	Eutectic β-$Nb_{2+x}C$–Nb *See also* section C–Nb in Table I-2.13	[2, 4–6, 67, 69]

(continued)

Table 4.18 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	Reference
NbC_{1-x}–Ni	–	1100	Formation of η-$Nb_4Ni_2C_{1-x}$, or η-$Nb_3Ni_3C_{1-x}$, or η-$Nb_6Ni_6C_{1-x}$	[1, 5, 67, 69, 81, 443–448, 554, 802, 809, 812–816, 827, 996, 999, 1069, 1070, 1085]
	Vacuum	1120	Formation of new phases initiates in the contact zone between the bulk solid components	
	–	1200	The max. solid solubility of Ni in near-stoichiometric NbC_{1-x} is ~2–3 mol.%; it increases with growth of C deficit in a carbide phase and inhibits order-disorder phase transformations (stabilizes NaCl structure in a carbide)	
	–	1250	The max. solid solubility of NbC_{1-x} in metallic Ni is ~2–3 mol.%	
	–	1300–1350	Metallic Ni on NbC_{1-x} plate instantaneously penetrates into the carbide and forms a eutectic (the process is identical in the case of both sintered and fused carbide materials)	
	–	~1320–1350	Eutectic NbC_{1-x}–Ni; the max. solid solubility of NbC_{1-x} in Ni is ~2–3 mol.% and that of Ni in NbC_{1-x} is ~9 mol.%	
	–	1325–1335	Formation of a liquid phase in the $NbC_{0.97}$–Ni powdered mixture with mean grain size ~50 μm	
	–	~1400	The max. solubility of NbC_{1-x} in molten Ni is 4.0 mol.%	
			See also Table 4.21	
			See also section C–Nb–Ni in Table I-2.14	
NbC_{1-x}–β-$Mo_{2\pm x}C$–TiC_{1-x}–δ-$TiN_{1\pm x}$–Ni	–	1000–1300	Formation of $(Nb,Mo,Ti)(C,N)_{1-x}$ complex monocarbonitride (cubic) phase	[990, 1067]
	–	1480	Powdered $TiC_{0.75}N_{0.2.5}$ – 2–8 vol.% NbC_{1-x} – 7 vol.% β-$Mo_{2\pm x}C$ – 8 vol.% Ni compositions were heat-treated to form cermets	
NbC_{1-x}–TaC_{1-x}–Ni			*See* section TaC_{1-x}–NbC_{1-x}–Ni in Table 2.21	
			See also section C–Nb–Ni–Ta in Table I-2.14	
NbC_{1-x}–TaC_{1-x}–TiC_{1-x}–δ-$TiN_{1\pm x}$–Ni			*See* section TaC_{1-x}–NbC_{1-x}–TiC_{1-x}–δ-$TiN_{1\pm x}$–Ni in Table 2.21	
NbC_{1-x}–TiC_{1-x}–Ni	–	–	The cast alloys crystallize in a metastable state, when the carbide phase does not decompose into two solid solutions even after long-term annealing	[812–814, 943, 1003, 1008, 1009, 1069, 1070, 1075]
	Vacuum, 0.01 Pa	1450	The interaction of $Nb_{0.24}Ti_{0.76}C_{0.96}$ phase with Ni melt leads to the dissolution of double carbide (preferentially C and Nb are transfered into the melt), phase separation (stratification) and formation of core-rim microstructures; Nb atoms diffuses from the core to the rim phase (~$Nb_{0.40}Ti_{0.60}C_{1-x}$)	

(continued)

Table 4.18 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	Reference
NbC_{1-x}–TiC_{1-x}–δ-$TiN_{1\pm x}$–Ni	Vacuum, 0.01 Pa	1450	The interaction of complex carbonitrides $Nb_{0.03–0.40}Ti_{0.60–0.97}C_{0.35–0.60}N_{0.40–0.65}$ with molten Ni leads to the incongruent dissolution of the components, preferentially – C and Nb; after the cooling metallic melts have a hypoeutectic structure with the primary precipitates of Ni based phase and intermetallide $Ni_{3\pm x}Nb$ (carbonitride constituents of eutectics is enriched with Nb)	[953, 993–996, 1000, 1068, 1073]
	Vacuum	1500–1510	Formation of $(Ti,Nb)(C,N)_{1-x}$–$Ti(C,N)_{1-x}$–Ni cermets with core-rim microstructures, the absent of the core can be caused by the high lattice mismatch at the core – inner rim interface; addition of NbC_{1-x} leads to the enhancement of wettability due to diffusion (dissolution) of Nb atoms into Ni phase	
NbC_{1-x}–TiC_{1-x}–δ-$TiN_{1\pm x}$–$TiO_{1\pm x}$–Ni	Vacuum, 0.1 Pa	1450	The interaction of $Nb_{0.40}Ti_{0.60}C_{0.50}N_{0.40}O_{0.10}$ oxycarbonitride phase with molten Ni is considered	[953]
NbC_{1-x}–TiC_{1-x}–δ-$TiN_{1\pm x}$–δ-$WC_{1\pm x}$–Ni	–	1300	Formation of $(Nb,W,Ti)(C,N)_{1-x}$ complex monocarbonitride (cubic) phase based cermet system	[990]
NbC_{1-x}–TiC_{1-x}–VC_{1-x}–δ-$WC_{1\pm x}$–Ni–Cr	Vacuum, ~1 Pa	1350–1450	Formation of $(Nb,Ti,V,W,Cr)C_{1-x}$ complex monocarbide (cubic) phases with core-rim microstructures, including elementary carbides, $(Nb,Ti)C_{1-x}$ binary monocarbide phase and Ni-based metallic alloy (binder)	[998]
NbC_{1-x}–VC_{1-x}–Ni	–	~1285–1315	Eutectic $(Nb,V)C_{1-x}$ (x = 0.1)–$(V,Nb)C_{1-x}$ ($x \approx 0.1$)–Ni; the max. solubility of NbC_{1-x} in VC_{1-x} is ~15 mol.% and in Ni is ~3 mol.%, and that of VC_{1-x} in NbC_{1-x} is ~20 mol.% and in Ni is ~3 mol.%, and that of Ni in NbC_{1-x} is ~9 mol.% and in VC_{1-x} is ~4 mol.%	[446]
			See also section C–Nb–Ni–V in Table I-2.14	
NbC_{1-x}–ZrC_{1-x}–Ni	–	~1290–1330	Eutectic $(Zr,Nb)C_{1-x}$–Ni (in the cast state due to the probable formation of a metastable ternary eutectic the alloys begin to melt at 80–100 °C lower than the equilibrium values); the max. solubility of $(Zr,Nb)C_{1-x}$ in Ni is ~2–3 mol.% and that of Ni in $(Zr,Nb)C_{1-x}$ is ~6 mol.%	[448, 1017]
			See also section C–Nb–Ni–Zr in Table I-2.14	
β-$Nb_{2+x}C$–Ni			*See* section C–Nb–Ni in Table I-2.14	
NbC_{1-x}–Ni–Cr	–	–	Porous NbC_{1-x} is impregnated perfectly with the Ni-9.5% Cr alloy	[1]
			See also sections C–Cr–Nb and C–Nb–Ni in Table I-2.14	

(continued)

Table 4.18 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	Reference
NbC_{1-x}–Os	–	1500	The solubility of Os in NbC_{1-x} is low *See also* section C–Nb–Os in Table I-2.14	[69]
β-$Nb_{2+x}C$–Os			*See* section C–Nb–Os in Table I-2.14	
NbC_{1-x}–P	–	–	Formation of Nb_2PC ($M_{n+1}AX_n$-phase) *See also* section C–Nb–P in Table I-2.14	[421, 422]
NbC_{1-x}–Pb			*See* Table 4.21	
NbC_{1-x}–Pd	–	1200–1500	The formation of (Pd,Nb) solid solutions and failure of NbC_{1-x} are occurred (8 h exposure); NbC_{1-x} is in equilibrium with ($Pd_{0.85}Nb_{0.15}$) alloy *See also* section C–Nb–Pd in Table I-2.14	[1, 5, 69, 449]
β-$Nb_{2+x}C$–Pd			*See* section C–Nb–Pd in Table I-2.14	
NbC_{1-x}–Pt	–	1200–1500	The formation of (Pt,Nb) solid solutions and the failure of NbC_{1-x} are occurred (8 h exposure); NbC_{1-x} is in equilibrium with α-$Pt_{3\pm x}Nb$ *See also* section C–Nb–Pt in Table I-2.14	[1, 5, 69, 449]
β-$Nb_{2+x}C$–Pt			*See* section C–Nb–Pt in Table I-2.14	
NbC_{1-x}–Pu			*See* section C–Nb–Pu in Table I-2.14	
β-$Nb_{2+x}C$–Pu			*See* section C–Nb–Pu in Table I-2.14	
NbC_{1-x}–Re	–	2000–2100	The solubility of Re in carbide is ~1.5 at.%; non-stoichiometric carbides are in equilibrium with χ-$NbRe_{3\pm x}$	[1, 69, 450, 451, 468–470, 570, 951]
	–	–	Stable and compatible with each other as cermet components	
	–	2210–2240	Eutectic NbC_{1-x}–Re; the max. solid solubility of Re in NbC_{1-x} is ~1.5–2.0 mol.% and that of NbC_{1-x} in Re is negligible too	
	He	2500	No interaction in the contact zone (1 h exposure) between the compact dense materials was observed (?) *See also* section C–Nb–Re in Table I-2.14	
β-$Nb_{2+x}C$–Re	–	1800–2000	The mutual solid solubilities of the components are very low *See also* section C–Nb–Re in Table I-2.14	[69, 450]
NbC_{1-x}–Rh	–	≥1500	The interaction results in the formation of $Rh_{3\pm x}Nb$ and C (graphite) *See also* section C–Nb–Rh in Table I-2.14	[69]
β-$Nb_{2+x}C$–Rh			*See* section C–Nb–Rh in Table I-2.14	
NbC_{1-x}–Ru	–	~1600	The interaction of non-stoichiometric NbC_{1-x} results in the formation of $NbRu_3C_{1-x}$ ($x \approx 0.6$); quasi-stoichiometric phases are in equilibrium with Ru *See also* section C–Nb–Ru in Table I-2.14	[69]

(continued)

Table 4.18 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	Reference
β-$Nb_{2+x}C$–Ru			*See* section C–Nb–Ru in Table I-2.14	
NbC_{1-x}–S	–	–	Formation of Nb_2SC_{1-x} (x = 0.6, $M_{n+1}AX_n$-phase), α/β-Nb_2CS_2 (?) *See also* section C–Nb–S in Table I-2.14	[421, 422]
NbC_{1-x}–Sb			*See* Table 4.21	
NbC_{1-x}–Si	Vacuum, $<10^{-4}$ Pa	600–700	No interdiffusion; no silicides or any other chemical species are formed (up to 1 h exposure) in the multilayers system of components	[5, 145, 421, 422, 452–455, 575, 584, 700, 701]
	Vacuum, 1.3 Pa	1100	Interaction in powdered equimolar mixture $NbC_{0.98}$ + Si (1 h exposure) leads to the formation of 4.7% $NbSi_2$ and 0.7% β-SiC	
	Vacuum, 10^{-3} Pa	1300	Under pressure (1 MPa) two-phase ($NbSi_2$ +SiC) layer is formed with silicide matrix and carbide discontinuous particles in the contact zone between the bulk components	
	Vacuum, 1.3 Pa	1300	Interaction in powdered equimolar mixture $NbC_{0.98}$ + Si (1 h exposure) leads to the formation of $NbSi_2$ (28.5%) and β-SiC (6.6%)	
	–	>1300	Formation of SiC and Nb silicides (or ternary phases, including $M_{n+1}AX_n$-phases, ?)	
	Vacuum, 1.3 Pa	1500	Interaction in powdered equimolar mixture $NbC_{0.98}$ + Si (1 h exposure) leads to the formation of 35.7% $NbSi_2$ and 10.2% β-SiC	
	Vacuum, 1.3 Pa	1700–1900	Interaction in equimolar powdered mixture $NbC_{0.98}$ + Si (1 h exposure) leads to the formation of 16–31% $NbSi_2$ and 5–10% β-SiC *See also* Table 4.21 *See also* section C–Nb–Si in Table I-2.14	
β-$Nb_{2+x}C$–Si			*See* section C–Nb–Si in Table I-2.14	
NbC_{1-x}–Sn	–	–	Formation of $Nb_{2\pm x}Sn_{1\pm y}C_{1-z}$ ($M_{n+1}AX_n$-phase with homogeneity range from $Nb_{2.20}Sn_{0.80}C_{0.93}$ to $Nb_{1.92}Sn_{1.08}C_{0.91}$); the solubility of Sn in NbC_{1-x} is extremely low *See also* Table 4.21 *See also* section C–Nb–Sn in Table I-2.14	[421, 422, 456, 457]
β-$Nb_{2+x}C$–Sn	–	–	The solubility of Sn in β-$Nb_{2+x}C$ is extremely low *See also* section C–Nb–Sn in Table I-2.14	[457]

(continued)

Table 4.18 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	Reference
NbC_{1-x}–Ta	–	≤1600	No interaction in the contact zone (2–5 h exposure) between the materials	[1, 5, 82, 145, 170, 442, 458, 463]
	–	1700	Under pressure 5 MPa the diffusion welded joint between NbC_{1-x} and Ta is produced (5–15 min exposure)	
	Vacuum	1700–2200	Formation of Ta–C, $Ta_{2+x}C$–$Nb_{2+x}C$ and TaC_{1-x}–NbC_{1-x} solid solutions in the contact zone between the bulk materials *See also* section C–Nb–Ta in Table I-2.14	
β-$Nb_{2+x}C$–Ta			*See* section C–Nb–Ta in Table I-2.14	
NbC_{1-x}–Tc			*See* section C–Nb–Tc in Table I-2.14	
NbC_{1-x}–Th			*See* section C–Nb–Th in Table I-2.14	
β-$Nb_{2+x}C$–Th			*See* section C–Nb–Th in Table I-2.14	
NbC_{1-x}–α/β-Ti	Vacuum, $<7 \times 10^{-4}$ Pa	1750	Molten Ti reacts extensively with NbC_{1-x} crucibles resulting in the formation of an alloy, which soaks into carbide materials; a new phase is apparently formed as a distinct layer based on TiC_{1-x} between carbide and metal Ti *See also* section C–Nb–Ti in Table I-2.14	[689]
β-$Nb_{2+x}C$–α/β-Ti	–	≤2280	The max. solid solubility of Ti varies from ~5 to ~9 at.% *See also* section C–Nb–Ti in Table I-2.14	[46, 69, 145, 459]
γ-$Nb_{2+x}C$–α/β-Ti	–	2600	The max. solid solubility of Ti in γ-$Nb_{2\pm x}C$ is ~1 at.% *See also* section C–Nb–Ti in Table I-2.14	[46, 69, 145, 459]
NbC_{1-x}–Tl			*See* Table 4.21	
NbC_{1-x}–U			*See* section C–Nb–U in Table I-2.14	
β-$Nb_{2+x}C$–U			*See* section C–Nb–U in Table I-2.14	
NbC_{1-x}–V			*See* section C–Nb–V in Table I-2.14	
β-$Nb_{2+x}C$–V			*See* section C–Nb–V in Table I-2.14	
NbC_{1-x}–W	–	1800	Under pressure 5 MPa the diffusion welded joint between NbC_{1-x} and W parts is produced (5–15 min exposure)	[1, 5, 67, 82, 170, 451, 458, 461, 465–467, 697]
	–	2000	The weak interaction between NbC_{1-x} and metallic W with formation of δ-$WC_{1\pm x}$	
	Vacuum	2200	The weak interaction between powdered NbC_{1-x} and metallic W (2 h exposure) with the formation of β-$(Nb,W)_{2+x}C$; practically, no interaction (exposure 5 h) between the bulk dense materials	
	Vacuum, 0.01 Pa	2200	The initiation of reaction between sintered dense NbC_{1-x} (x = 0.11–0.27) and compact metallic W (exposure 2 h)	
	Vacuum, 0.01 Pa	2300	The initiation of reaction between sintered dense NbC_{1-x} ($x \approx 0$) and compact metallic W (exposure 2 h)	
	Vacuum	2500	No interaction in the contact zone (4 h exposure) between the compact materials (?) *See also* section C–Nb–W in Table I-2.14	

(continued)

Table 4.18 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	Reference
β-$Nb_{2+x}C$–W	–	2000	The max. solid solubility of W in β-$Nb_{2+x}C$ is ~8 at.% *See also* section C–Nb–W in Table I-2.14	[69, 145, 460, 461]
γ-$Nb_{2\pm x}C$–W	– –	2500 2690	The max. solid solubility of W is ~8 at.% The max. solid solubility of W is ~2 at.% *See also* section C–Nb–W in Table I-2.14	[69, 145, 460, 461]
NbC_{1-x}–Zn	–	–	Formation of related to η-carbide type Nb_2ZnC_{1-x} phase	[81, 563]
NbC_{1-x}–α/β-Zr	Vacuum	750–1500	Interaction in the powdered mixtures (mean grain size – 1–2 μm) of components leads to the formation of highly non-stoichiometric $ZrC_{\sim 0.6}$ phase *See also* section C–Nb–Zr in Table I-2.14	[816, 827]
β-$Nb_{2+x}C$–α/β-Zr	–	–	The solubility of Zr in β-$Nb_{2+x}C$ is low *See also* section C–Nb–Zr in Table I-2.14	[69, 145, 462]

[a]The parameters of wettability of niobium monocarbide phases by liquid metals at various temperatures are listed in Table 4.21

Table 4.19 Chemical interaction and/or compatibility of niobium carbide phases with refractory compounds at elevated, high and ultra-high temperatures (reaction systems are given mainly in alphabetical order)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	Reference
NbC_{1-x}–α-Al_2O_3	CO_2, 1.4×10^{-15} Pa	1000	Calculated equilibrium pressure of the interaction between the components	[576–580, 890]
	Ar flow	1650–1750	No chemical interaction between the powdered components; no intermediate (amorphous or crystalline) phases at the intergranular interfaces	
α-Nb_2C–α-Al_2O_3	CO_2, 2.2×10^{-18} Pa	1000	Calculated equilibrium pressure of the interaction between the components	[576, 577]
NbC_{1-x}–Al_4C_3	–	Up to ~2000	No mutual solid solubilities	[417–419, 559, 561, 569, 589, 900, 901, 948]
	–	–	Formation of Nb_2AlC, α/β-Nb_4AlC_3 ($M_{n+1}AX_n$-phases) *See also* section NbC_{1-x}–Al in Table 4.18 *See also* section C–Al–Nb in Table I-2.14	
NbC_{1-x}–Al_4C_3–SiC–TiC_{1-x}	–	–	Formation of solid solutions $(Ti_{1-x}Nb_x)_3(Si_{1-x}Al_x)C_2$ based on Ti_3SiC_2 *See also* section NbC_{1-x}–SiC–TiC_{1-x}–Al in Table 4.18	[903]

(continued)

Table 4.19 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	Reference
NbC_{1-x}–Al_4C_3–TiC_{1-x}	–	–	Formation of $(Nb,Ti)_2AlC$, $(Nb,Ti)_4AlC_3$ and $(Nb,Ti)_5AlC_4$ ($M_{n+1}AX_n$-phase solid solutions); $(Nb_{0.5}Ti_{0.5})_2AlC$, $(Nb_{0.5}Ti_{0.5})_5AlC_4$ and $(Nb_{1-x}Ti_x)_4AlC_3$ ($0 \leq x \leq 0.3$) were synthesized *See also* section NbC_{1-x}–TiC_{1-x}–Al in Table 4.18 *See also* section C–Al–Nb–Ti in Table I-2.14	[421, 422, 561, 573, 912–914, 954]
NbC_{1-x}–Al_4C_3–VC_{1-x}	–	–	Formation of $(Nb,V)_2AlC$ and $(Nb,V)_4AlC_3$ ($M_{n+1}AX_n$-phase solid solutions); compositions $(Nb_{0.5}V_{0.5})_2AlC$ and $(Nb_{0.5}V_{0.5})_4AlC_3$ were synthesized *See also* section NbC_{1-x}–VC_{1-x}–Al in Table 4.18 *See also* section C–Al–Nb–V in Table I-2.14	[421, 422, 573, 912, 1016]
NbC_{1-x}–Al_4C_3–ZrC_{1-x}	–	–	Formation of $(Nb,Zr)_2AlC$ ($M_{n+1}AX_n$-phase solid solution); $(Nb_{0.6}Zr_{0.4})_2AlC$ and $(Nb_{0.8}Zr_{0.2})_2AlC$ were synthesized *See also* section NbC_{1-x}–ZrC_{1-x}–Al in Table 4.18 *See also* section C–Al–Nb–Zr in Table I-2.14	[419, 421, 422, 573, 912, 1014–1016]
NbC_{1-x}–$B_{4\pm x}C$	–	>∼1500	Formation of $NbB_{2\pm x}$ and C *See also* section C–B–Nb in Table I-2.14	[5, 67, 472–474]
NbC_{1-x}–$B_{4\pm x}C$–α-BN	–	≤2100	No contact reaction between powdered NbC_{1-x} and compact dense $B_{4\pm x}C$–α-BN composition *See also* section C–B–N–Nb in Table I-2.14	[804]
β-$Nb_{2+x}C$–$B_{4\pm x}C$			*See* section C–B–Nb in Table I-2.14	
NbC_{1-x}–α-BN	–	≥2250	Formation of $NbB_{2\pm x}$ *See also* section C–B–N–Nb in Table I-2.14	[5, 475]
NbC_{1-x}–$CeN_{1\pm x}$	–	–	Terminal mutual solid solubilities between the components, ?	[5]
NbC_{1-x}–$CeP_{1\pm x}$	–	–	Terminal mutual solid solubilities between the components, ?	[5]
NbC_{1-x}–$CeS_{1\pm x}$	–	–	Terminal mutual solid solubilities between the components, ?	[5]
NbC_{1-x}–Cr_3C_{2-x}	–	1740–1820	The max. solubility of "imaginary" phase 'CrC_{1-x}' in NbC_{1-x} is ∼26–29 mol.%; NbC_{1-x} phase is in equilibrium with C (graphite) and Cr carbides *See also* section C–Cr–Nb in Table I-2.14	[431, 585, 588, 932]

(continued)

Table 4.19 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	Reference
NbC_{1-x}–$Cr_7C_{3\pm x}$	–	1050–1750	The max. solubility of "imaginary" phase 'CrC_{1-x}' in NbC_{1-x} is ~25–28 mol.%; NbC_{1-x} phase is in equilibrium with all Cr carbides and intermetallide and metallic Cr	[52, 69, 429, 431, 553, 585, 588, 932]
	–	1580–1610	The max. solubility of NbC_{1-x} in $Cr_7C_{3\pm x}$ is corresp. to compos. ~$(Cr_{>0.99}Nb_{<0.006})_7C_{3\pm x}$	
	–	~1685	Eutectic NbC_{1-x}–$Cr_7C_{3\pm x}$; the max. solubility of $Cr_7C_{3\pm x}$ in NbC_{1-x} is corresponding to composition ~$(Nb_{0.90}Cr_{0.10})C_{1-x}$ and that of NbC_{1-x} in $Cr_7C_{3\pm x}$ – to composition ~$(Cr_{0.99}Nb_{0.01})_7C_{3\pm x}$ *See also* section C–Cr–Nb in Table I-2.14	
NbC_{1-x}–$Cr_{23}C_{6\pm x}$	–	1550	The max. solubility of NbC_{1-x} in $Cr_{23}C_{6\pm x}$ is corresponding to composition ~$(Cr_{0.98-0.99}Nb_{0.01-0.02})_{23}C_{6\pm x}$ *See also* section C–Cr–Nb in Table I-2.14	[431, 585–588, 932]
β-$Nb_{2+x}C$–Cr_3C_{2-x}, $Cr_7C_{3\pm x}$, $Cr_{23}C_{6\pm x}$	–	1600–1680	The max. solubility of Cr carbides in β-$Nb_{2+x}C$ is corresp. to composition ~$(Nb_{0.95-0.99}Cr_{0.01-0.05})_{2+x}C$; β-$Nb_{2+x}C$ phase is in equilibrium with λ_1/λ_2-$NbCr_{2\pm x}$, NbC_{1-x} and metallic Nb *See also* section C–Cr–Nb in Table I-2.14	[431, 585, 932]
NbC_{1-x}–Cr_2O_3	CO_2, 1.2×10^{-6} Pa	1000	Calculated equilibrium pressure of the interaction between the components	[576, 577]
α-Nb_2C–Cr_2O_3	CO_2, 1.8×10^{-9} Pa	1000	Calculated equilibrium pressure of the interaction between the components	[576, 577]
NbC_{1-x}–$DyN_{1\pm x}$	–	–	Monocarbonitride (cubic) continuous solid solution (complete solubility in the system?)	[5]
NbC_{1-x}–$ErN_{1\pm x}$	–	–	Monocarbonitride (cubic) continuous solid solution (complete solubility in the system?)	[5]
NbC_{1-x}–$EuN_{1\pm x}$	–	–	Terminal mutual solid solubilities between the components (?)	[5]
NbC_{1-x}–EuO	–	–	Terminal mutual solid solubilities between the components (?)	[5]
NbC_{1-x}–Fe_3C	–	–	Not (or very slightly) soluble in each other in solid state *See also* section C–Fe–Nb in Table I-2.14	[69, 81]
NbC_{1-x}–Fe_2O_3	CO_2, 1.2 Pa	1000	Calculated equilibrium pressure of the interaction between the components	[576, 577]
α-Nb_2C–Fe_2O_3	CO_2, 1.8×10^{-3} Pa	1000	Calculated equilibrium pressure of the interaction between the components	[576, 577]
NbC_{1-x}–$Fe_{3+x}Al$	Vacuum, 4 Pa	1420	The liquid-phase sintering of powdered $NbC_{\sim 1.0}$ (mean particle size – 1.1 μm) – 7 mol.% $Fe_{3+x}Al$ (mean particle size <40 μm) composition leads to the formation of NbC_{1-x} – (Fe,Al,C) hard alloy (cermet) system	[1011]

(continued)

Table 4.19 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	Reference
NbC_{1-x}–$GdN_{1\pm x}$	–	–	Terminal mutual solid solubilities between the components (?)	[5]
NbC_{1-x}–HfC_{1-x}			*See* section HfC_{1-x}–NbC_{1-x} in Table 3.20 *See also* section C–Hf–Nb in Table I-2.14	
β-$Nb_{2+x}C$–HfC_{1-x}			*See* section HfC_{1-x}–β-$Nb_{2+x}C$ in Table 3.20 *See also* section C–Hf–Nb in Table I-2.14	
NbC_{1-x}–HfC_{1-x}–TaC_{1-x}			*See* section TaC_{1-x}–HfC_{1-x}–NbC_{1-x} in Table 2.22	
NbC_{1-x}–HfC_{1-x}–$ThC_{1\pm x}$			*See* section HfC_{1-x}–NbC_{1-x}–$ThC_{1\pm x}$ in Table 3.20	
NbC_{1-x}–HfC_{1-x}–TiC_{1-x}			*See* section HfC_{1-x}–NbC_{1-x}–TiC_{1-x} in Table 3.20	
NbC_{1-x}–HfC_{1-x}–$UC_{1\pm x}$			*See* section HfC_{1-x}–NbC_{1-x}–$UC_{1\pm x}$ in Table 3.20 *See also* section C–Hf–Nb–U in Table I-2.14	
NbC_{1-x}–HfC_{1-x}–VC_{1-x}			*See* section HfC_{1-x}–NbC_{1-x}–VC_{1-x} in Table 3.20 *See also* section C–Hf–Nb–V in Table I-2.14	
NbC_{1-x}–HfC_{1-x}–δ-$WC_{1\pm x}$			*See* section HfC_{1-x}–NbC_{1-x}–δ-$WC_{1\pm x}$ in Table 3.20	
NbC_{1-x}–HfC_{1-x}–ZrC_{1-x}			*See* section HfC_{1-x}–NbC_{1-x}–ZrC_{1-x} in Table 3.20	
NbC_{1-x}–δ-NbN_{1-x}–HfC_{1-x}			*See* section HfC_{1-x}–NbC_{1-x}–δ-NbN_{1-x} in Table 3.20	
NbC_{1-x}–$HfN_{1\pm x}$	–	1600–2200	Monocarbonitride (cubic) continuous solid solution (complete solubility in the system)	[5, 479]
NbC_{1-x}–$HoN_{1\pm x}$	–	–	Monocarbonitride (cubic) continuous solid solution (complete solubility in the system?)	[5]
NbC_{1-x}–$LaN_{1\pm x}$	–	–	Terminal mutual solid solubilities between the components (?)	[5]
NbC_{1-x}–$LaB_{6\pm x}$	Ar	1700	The powdered components interact with the formation of $NbB_{2\pm x}$ and C; complete decomposition of carbide (6 h exposure)	[911]
NbC_{1-x}–$LaP_{1\pm x}$	–	–	Terminal mutual solid solubilities between the components (?)	[5]
NbC_{1-x}–LaS_{1-x}	–	–	Terminal mutual solid solubilities between the components (?)	[5]
NbC_{1-x}–$LuN_{1\pm x}$	–	–	Monocarbonitride (cubic) continuous solid solution (complete solubility in the system?)	[5]

(continued)

Table 4.19 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	Reference
NbC_{1-x}–MgO	Vacuum 0.01-0.1 Pa	1300	The powdered mixtures of the components lose mass with the release of volatile products (metallic Mg and CO) and formation C-depleted NbC_{1-x} phases	[1, 5, 82, 480, 481, 550]
	Vacuum	1800–2300	Interaction with the formation of metallic Nb and NbC_xO_y oxycarbide phases in the contact zone of compact bulk materials	
			See also section C–Mg–Nb–O in Table I-2.14	
NbC_{1-x}–Mn_5C_2, Mn_7C_3	–	700–1000	Not (or very slightly) soluble in each other at solid state; NbC_{1-x} phase is in equilibrium with all Mn carbides and intermetallide and metallic Mn	[81, 585]
			See also section C–Mn–Nb in Table I-2.14	
NbC_{1-x}–α-MoC_{1-x}	–	1900–2000	Extended monocarbide solid solution based on NbC_{1-x} – up to composition $\sim Nb_{0.10–0.15}Mo_{0.85–0.90}C_{1-x}$; the max. solubility of Nb in α-MoC_{1-x} is ~2 at.%	[5, 54, 57, 69, 81, 440, 441, 482, 568, 946,
	–	~2000–2600	Monocarbide (cubic) continuous solid solution	988]
			See also section C–Mo–Nb in Table I-2.14	
NbC_{1-x}–α-MoC_{1-x}–$UC_{1\pm x}$	–	2000	Extended monocarbide (cubic) solid solution based on NbC_{1-x}–$UC_{1\pm x}$ continuous solid solution; the solubility of α-MoC_{1-x} in NbC_{1-x} is >90 mol.% (the solubilities of NbC_{1-x} and $UC_{1\pm x}$ in α-MoC_{1-x} are ~1–3 mol.%)	[484]
			See also section C–Mo–Nb–U in Table I-2.14	
NbC_{1-x}–α-MoC_{1-x}–ZrC_{1-x}	–	1900–2000	Extended monocarbide (cubic) solid solution based on NbC_{1-x}–ZrC_{1-x} continuous solid solution; the solubility of α-MoC_{1-x} in NbC_{1-x} and ZrC_{1-x} are >90 mol.% and >80 mol.%, respectively (the solubilities of NbC_{1-x} and ZrC_{1-x} in α-MoC_{1-x} are low)	[27, 484, 759]
	–	>2000	Monocarbide (cubic) continuous solid solution (complete solubility in the system)	
			See also section C–Mo–Nb–Zr in Table I-2.14	
NbC_{1-x}–β-$Mo_{2\pm x}C$	–	~2000	The max. solubility of Nb in β-$Mo_{2\pm x}C$ corresponds to ~$(Mo_{0.6}Nb_{0.4})_{2\pm x}C$ composition	[54, 57, 69, 441, 946]
			See also section C–Mo–Nb in Table I-2.14	
α-Nb_2C–α-$Mo_{2+x}C$	–	1200	Semicarbide (orthorhombic) solid solution based on α-Nb_2C	[562]
			See also section C–Mo–Nb in Table I-2.14	
β-$Nb_{2+x}C$–β-$Mo_{2\pm x}C$	–	1750	Semicarbide (hexagonal) solid solution based on β-$Nb_{2+x}C$; the max. solubility of Mo is ~3 at.%	[54, 57, 69, 441, 562]
			See also section C–Mo–Nb in Table I-2.14	

(continued)

Table 4.19 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	Reference
β-$Nb_{2+x}C$–β-$Mo_{2\pm x}C$–α-$Ta_{2+x}C$			*See* section α-$Ta_{2+x}C$–β-$Mo_{2\pm x}C$–β-$Nb_{2+x}C$ in Table 2.22 *See also* section C–Mo–Nb–Ta in Table I-2.14	
NbC_{1-x}–δ-NbN_{1-x}	–	1250–2130	Monocarbonitride (cubic) continuous solid solution (the variation of lattice parameter a, nm with composition for $Nb(C_{1-x}N_x)$ phases ($0 \leq x \leq 1$) is linear: $a = 0.4470 - 0.0080x$) *See also* section C–N–Nb in Table I-2.14	[5, 34, 69, 485–487, 494, 495, 508, 842, 899, 1030, 1090–1092]
β-$Nb_{2+x}C$–β-$Nb_{2\pm x}N$	–	1250–1450	Semicarbonitride continuous solid solution *See also* section C–N–Nb in Table I-2.14	[69, 485]
NbC_{1-x}–δ-NbN_{1-x}–NbO	–	–	Formation of $NbC_{1-x}N_yO_z$ oxycarbonitride phase[a] based on NbC_{1-x}–δ-NbN_{1-x} (cubic) continuous solid solution	[34, 59]
NbC_{1-x}–δ-NbN_{1-x}–δ-$WC_{1\pm x}$	–	–	General consideration of the system *See also* section C–N–Nb–W in Table I-2.14	[427, 428]
NbC_{1-x}–δ-NbN_{1-x}–$HfN_{1\pm x}$	Pure N_2	1200–2500	Monocarbonitride (cubic) continuous solid solution (complete solubility in the system)	[34]
NbC_{1-x}–δ-NbN_{1-x}–δ-$TiN_{1\pm x}$	Pure N_2	1200–2500	Monocarbonitride (cubic) continuous solid solution (complete solubility in the system)	[34, 1090–1092]
NbC_{1-x}–δ-NbN_{1-x}–δ-$TiN_{1\pm x}$–$ZrN_{1\pm x}$	Pure N_2	1200–2500	Monocarbonitride (cubic) continuous solid solution (complete solubility in the system)	[34]
NbC_{1-x}–δ-NbN_{1-x}–δ-VN_{1-x}	Pure N_2	1200–2500	Monocarbonitride (cubic) continuous solid solution (complete solubility in the system)	[34]
NbC_{1-x}–δ-NbN_{1-x}–$ZrN_{1\pm x}$	Pure N_2	1200–2500	Monocarbonitride (cubic) continuous solid solution (complete solubility in the system)	[34]
NbC_{1-x}–$NbB_{2\pm x}$	–	$\leq$2000	The components are compatible and virtually insoluble in each other	[2, 5, 473, 474, 488]
	–	~2600–2900 (?)	Eutectic NbC_{1-x}–$NbB_{2\pm x}$; the max. solid solubility of $NbB_{2\pm x}$ in NbC_{1-x} is ~7 mol.% and that of NbC_{1-x} in $NbB_{2\pm x}$ is ~1–2 mol.% *See also* section C–B–Nb in Table I-2.14	
NbC_{1-x}–$NbB_{1\pm x}$	–	~2800	Eutectic NbC_{1-x}–$NbB_{1\pm x}$ *See also* section C–B–Nb in Table I-2.14	[5, 473, 474]
β-$Nb_{2+x}C$–$NbB_{1\pm x}$			*See* section C–B–Nb in Table I-2.14	

(continued)

Table 4.19 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	Reference
NbC_{1-x}–β-Nb_2O_5	Vacuum 10 Pa	1200–1500	Decrease of the content of carbon in NbC_{1-x} and formation of new phases (β-$Nb_{2+x}C$, traces NbO) in the powdered mixtures (1 h exposure)	[1, 59, 82, 481, 505–507, 550, 742–744, 868]
	CO, 0.01–101 kPa	1300–2000	Formation of $NbC_{1-x}O_y$ (NbC_{1-x}—'NbO' oxycarbide (cubic) solid solution) with approximate homogeneity limits: ~$NbC_{0.70}$–~$NbC_{0.50}O_{0.15}$–~$NbC_{0.35}O_{0.45}$–~$NbC_{0.35}O_{0.65}$–~$NbC_{0.99}$	
	Vacuum 10 Pa	1600–1900	Formation of β-$Nb_{2+x}C$ and metallic Nb (1 h exposure)	
	Vacuum, Ar	1900	Formation of oxycarbide (cubic) phase $NbC_{0.24–0.71}O_{0.05–0.62}$ ($0.65 \leq 1-x+y \leq 1.12$)	
	Vacuum <10 Pa	>1900	In the powdered mixtures with general O/C ratio = 1.01–1.05 the only product is metallic Nb with some O and C contaminations *See also* section C–Nb–O in Table I-2.14	
β-$Nb_{2+x}C$–β-Nb_2O_5	CO, 0.01–101 kPa	1300–2000	Formation of $Nb_2C_xO_y$ oxycarbide (hexagonal) solid solution based on β-$Nb_{2+x}C$ with approximate homogeneity composition limits: $Nb_{2.5}C$–$Nb_2(C_{0.3}O_{0.3})$–$Nb_2(C_{0.15}O_{0.7})$–$Nb_2(C_{0.35}O_{0.8})$–$Nb_{2.0}C$	[59, 742–744, 868]
	Vacuum, Ar	1900	Formation of oxycarbide (hexagonal) phase $Nb_2C_{0.33–0.34}O_{0.08–0.17}$ ($0.42 \leq x+y \leq 0.50$) *See also* section C–Nb–O in Table I-2.14	
NbC_{1-x}–NbO	Ar, 0.1 MPa	1400	Formation of oxycarbide (cubic) phases NbC_xO_y	[5, 743, 868]
	Vacuum, 0.13 Pa	1600–1700	Decomposition of oxycarbide (cubic) phases NbC_xO_y	
	–	–	Terminal mutual solid solubilities between the components, ? *See also* section C–Nb–O in Table I-2.14	
β-$Nb_{2+x}C$–NbO	Ar, 0.1 MPa	1400	Formation of oxycarbide (hexagonal) phases $Nb_2C_xO_y$	[743, 868]
	Vacuum, 0.13 Pa	1600–1700	Decomposition of oxycarbide (hexagonal) phases $Nb_2C_xO_y$ *See also* section C–Nb–O in Table I-2.14	
NbC_{1-x}–$NbSi_2$	–	~1880–1920	Eutectic (degenerated) NbC_{1-x}–$NbSi_2$; no mutual solubilities between the components *See also* section C–Nb–Si in Table I-2.14	[2, 5, 123, 452–454]
β-$Nb_{2+x}C$–$NbSi_2$			*See* section C–Nb–Si in Table I-2.14	
NbC_{1-x}–$NdN_{1\pm x}$	–	–	Terminal mutual solid solubilities between the components (?)	[5]

(continued)

Table 4.19 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	Reference
NbC_{1-x}–$Nd_2Fe_{14}B$	Pure Ar	1000	The solubility of NbC_{1-x} in complex boride phase is ~8 mol.%	[1007]
NbC_{1-x}–NpC_{1-x}	–	–	Terminal mutual solid solubilities between the components (?)	[5]
NbC_{1-x}–$NpN_{1\pm x}$	–	–	Monocarbonitride (cubic) continuous solid solution (complete solubility in the system?)	[5]
NbC_{1-x}–$PrN_{1\pm x}$	–	–	Terminal mutual solid solubilities between the components (?)	[5]
NbC_{1-x}–PuC_{1-x}	–	>1600	Monocarbide (cubic) continuous solid solution (complete solubility in the system) *See also* section C–Nb–Pu in Table I-2.14	[5, 69]
NbC_{1-x}–$PuN_{1\pm x}$	–	–	Terminal mutual solid solubilities between the components (?)	[5]
NbC_{1-x}–$PuP_{1\pm x}$	–	–	Terminal mutual solid solubilities between the components (?)	[5]
NbC_{1-x}–$PuS_{1\pm x}$	–	–	Terminal mutual solid solubilities between the components (?)	[5]
NbC_{1-x}–ScC_{1-x}	–	–	Monocarbide (cubic) continuous solid solution (complete solubility in the system)	[5, 556]
NbC_{1-x}–$ScN_{1\pm x}$	–	–	Monocarbonitride (cubic) continuous solid solution (complete solubility in the system?)	[5]
NbC_{1-x}–α/β-SiC	–	–	Compatible with each other within the temperature range of the thermal stabilities of the phases *See also* section C–Nb–Si in Table I-2.14	[5, 452–454]
β-$Nb_{2+x}C$–SiC			*See* section C–Nb–Si in Table I-2.14	
NbC_{1-x}–α/β-Si_3N_4	–	~1400–1600	Interaction leads to the formation of Nb silicide phases and SiC	[1005, 1006]
NbC_{1-x}–SiO_2–Al_2O_3–CaO–Fe_2O_3 (basalt)	–	1400	NbC_{1-x} is not stable in molten basalt (SiO_2 – 15% Al_2O_3 – 13% CaO – 13% Fe_2O_3) at short-term exposure (0.5–1 h) with average dissolution rate ~0.2 g cm^{-2} h^{-1}	[1027]
NbC_{1-x}–$SmN_{1\pm x}$	–	–	Terminal mutual solid solubilities between the components (?)	[5]
NbC_{1-x}–TaC_{1-x}			*See* section TaC_{1-x}–NbC_{1-x} in Table 2.22 *See also* section C–Nb–Ta in Table I-2.14	
NbC_{1-x}–TaC_{1-x}–$ThC_{1\pm x}$			*See* section TaC_{1-x}–NbC_{1-x}–$ThC_{1\pm x}$ in Table 2.22	
NbC_{1-x}–TaC_{1-x}–TiC_{1-x}			*See* section TaC_{1-x}–NbC_{1-x}–TiC_{1-x} in Table 2.22 *See also* section C–Nb–Ta–Ti in Table I-2.14	
NbC_{1-x}–TaC_{1-x}–δ-TaN_{1-x}–TiC_{1-x}			*See* section TaC_{1-x}–δ-TaN_{1-x}–NbC_{1-x}–TiC_{1-x} in Table 2.22	
NbC_{1-x}–TaC_{1-x}–$UC_{1\pm x}$			*See* section TaC_{1-x}–NbC_{1-x}–$UC_{1\pm x}$ in Table 2.22	

(continued)

Table 4.19 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	Reference
NbC_{1-x}–TaC_{1-x}–VC_{1-x}			*See* section TaC_{1-x}–NbC_{1-x}–VC_{1-x} in Table 2.22	
NbC_{1-x}–TaC_{1-x}–δ-$WC_{1\pm x}$			*See* section TaC_{1-x}–NbC_{1-x}–δ-$WC_{1\pm x}$ in Table 2.22 *See also* section C–Nb–Ta–W in Table I-2.14	
NbC_{1-x}–TaC_{1-x}–ZrC_{1-x}			*See* section TaC_{1-x}–NbC_{1-x}–ZrC_{1-x} in Table 2.22	
β-$Nb_{2+x}C$–α-$Ta_{2+x}C$			*See* section α-$Ta_{2+x}C$–β-$Nb_{2+x}C$ in Table 2.22 *See also* section C–Nb–Ta in Table I-2.14	
β-$Nb_{2+x}C$–α-$Ta_{2+x}C$–γ-$W_{2\pm x}C$			*See* section α-$Ta_{2+x}C$–β-$Nb_{2+x}C$–γ-$W_{2\pm x}C$ in Table 2.22 *See also* section C–Nb–Ta–W in Table I-2.14	
NbC_{1-x}–δ-TaN_{1-x}	N_2, 3 MPa	1800	Monocarbonitride (cubic) continuous solid solution (complete solubility in the system)	[42]
NbC_{1-x}–ε-TaN_{1-x}	N_2	–	The max. solid solubility of cubic NbC_{1-x} in hexagonal ε-TaN_{1-x} is much lower (practically negligible) than that of ε-TaN_{1-x} in NbC_{1-x}	[5, 159]
NbC_{1-x}–$TbN_{1\pm x}$	–	–	Monocarbonitride (cubic) continuous solid solution (complete solubility in the system?)	[5]
NbC_{1-x}–TcC_{1-x}	–	–	Monocarbide continuous solid solution (complete solubility in the system?) *See also* section C–Nb–Tc in Table I-2.14	[546, 931]
β-$Nb_{2+x}C$–TcC_{1-x}			The mutual solid solubilities of the components are low (?) *See also* section C–Nb–Tc in Table I-2.14	[546, 931]
NbC_{1-x}–$ThC_{1\pm x}$ (α-Th, γ-ThC_{2-x})	–	1500	Compatible with each other (mutual solid solubilities are very low) *See also* section C–Nb–Th in Table I-2.14	[5, 69, 490]
NbC_{1-x}–$ThC_{1\pm x}$–TiC_{1-x}	–	>2000	Practically, no mutual solubilities between NbC_{1-x}–TiC_{1-x} monocarbide continuous solid solution and $ThC_{1\pm x}$	[5, 477]
NbC_{1-x}–$ThC_{1\pm x}$–$UC_{1\pm x}$	–	>2000	Monocarbide continuous solid solution with great miscibility gap because of low mutual solubilities in the NbC_{1-x}–$ThC_{1\pm x}$ system	[5, 477]
NbC_{1-x}–$ThC_{1\pm x}$–VC_{1-x}	–	>2000	Practically, no mutual solubilities between NbC_{1-x}–VC_{1-x} monocarbide continuous solid solution and $ThC_{1\pm x}$	[5, 477]
NbC_{1-x}–$ThC_{1\pm x}$–ZrC_{1-x}	–	> 2000	Practically, no mutual solubilities between NbC_{1-x}–ZrC_{1-x} monocarbide (cubic) continuous solid solution and $ThC_{1\pm x}$	[5, 477]

(continued)

Table 4.19 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	Reference
β-$Nb_{2+x}C$–$ThC_{1\pm x}$	–	1500	Mutual solid solubilities are very low *See also* section C–Nb–Th in Table I-2.14	[69, 490]
NbC_{1-x}–γ-ThC_{2-x}	–	1500	Compatible with each other (mutual solid solubilities are very low) *See also* section C–Nb–Th in Table I-2.14	[5, 69, 490]
β-$Nb_{2+x}C$–γ-ThC_{2-x}			*See* section C–Nb–Th in Table I-2.14	
NbC_{1-x}–$ThN_{1\pm x}$	–	–	Extended monocarbonitride (cubic) solid solutions (terminal solubility?)	[5]
NbC_{1-x}–$ThP_{1\pm x}$	–	–	Terminal mutual solid solubilities between the components (?)	[5]
NbC_{1-x}–$ThS_{1\pm x}$	–	–	Terminal mutual solid solubilities between the components (?)	[5]
NbC_{1-x}–TiC_{1-x}	–	~0	Monocarbide extended solid solutions; the miscibility gap ranges from ~$Nb_{0.15}Ti_{0.85}C_{1-x}$ to ~$Nb_{0.7}Ti_{0.3}C_{1-x}$ (theoretically calculated)	[5, 46, 69, 145, 332, 459, 491, 492, 571, 572, 581, 817, 856, 876, 897, 920, 946, 955, 973–975, 988, 1001, 1002, 1030, 1087, 1093, 1094, 1100]
	–	~50–3100	Monocarbide continuous solid solution (the variation of lattice parameter a, nm with composition for $(Nb_{1-y}Ti_y)C_{1-x}$ phases ($x \approx 0$, $0 \le y \le 1$) is linear: $a = 0.4470 - 0.0141y$); critical point of the miscibility gap is corresponding to ~50 °C and ~$Nb_{0.40}Ti_{0.60}C_{1-x}$ composition (theoretically calculated)	
	–	1700–2000	No complete homogenization is achieved (1 h exposure) for two-phase carbide mixture prepared by carbothermal reduction of metal oxides	
	–	–	Formation of $(Nb_{0.8}Ti_{0.2})_4C_3T_x$ (MXene, 2D-Nb_4C_3-based solid solution, where T_x is a surface termination) *See also* section C–Nb–Ti in Table I-2.14	
β-$Nb_{2+x}C$–TiC_{1-x}	–	≤2280	The max. solubility of Ti in β-$Nb_{2+x}C$ varies from ~5 to ~9 at.% *See also* section C–Nb–Ti in Table I-2.14	[69, 145, 459, 492]
γ-$Nb_{2\pm x}C$–TiC_{1-x}	–	2600	The max. solubility of Ti in γ-$Nb_{2\pm x}C$ is ~1 at.% (?) *See also* section C–Nb–Ti in Table I-2.14	[69, 145, 459, 492]
NbC_{1-x}–δ-NbN_{1-x}–TiC_{1-x}	Pure N_2	1200–2500	Monocarbonitride (cubic) continuous solid solution (complete solubility in the system)	[34, 571, 572, 1030]
NbC_{1-x}–δ-NbN_{1-x}–TiC_{1-x}–δ-$TiN_{1\pm x}$	–	1000–1500	Monocarbonitride (cubic) continuous solid solution (complete solubility in the system, the variation of lattice parameter a, nm with composition for $(Nb_{1-y}Ti_y)(C_{1-x}N_x)$ phases ($0 \le x \le 1$, $0 \le y \le 1$): $a = 0.4470 - 0.0080x - 0.0141y - 0.0002xy$)	[1, 494, 571, 572, 899, 997, 1030, 1071]

(continued)

Table 4.19 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	Reference
NbC_{1-x}–TiC_{1-x}–$UC_{1\pm x}$	–	2000–2050	Monocarbide continuous solid solution with great miscibility gap because of low mutual solubilities in the TiC_{1-x}–$UC_{1\pm x}$ system *See also* section C–Nb–Ti–U in Table I-2.14	[5, 69, 477]
NbC_{1-x}–TiC_{1-x}–VC_{1-x}	–	1000	Monocarbide continuous solid solution with the miscibility gap as the mutual solubilities in the NbC_{1-x}–VC_{1-x} system are ~10 mol.%	[5, 81, 435, 477, 493, 494, 817, 899]
	–	1600–2000	Monocarbide (cubic) continuous solid solution (complete solubility in the system) *See also* section C–Nb–Ti–V in Table I-2.14	
NbC_{1-x}–TiC_{1-x}–δ-$WC_{1\pm x}$	–	1450	Extended solid solution based on NbC_{1-x}–TiC_{1-x} monocarbide continuous solid solution; the solubilities of δ-$WC_{1\pm x}$ in NbC_{1-x} and TiC_{1-x} are ~10 and ~35 mol.%, respectively (the solubilities of NbC_{1-x} and TiC_{1-x} in δ-$WC_{1\pm x}$ are very low)	[5, 81, 145, 477, 489]
	–	1900–2000	Extended solid solution based on NbC_{1-x}–TiC_{1-x} monocarbide continuous solid solution; the solubilities of δ-$WC_{1\pm x}$ in NbC_{1-x} and TiC_{1-x} are ~35 and ~60 mol.%, respectively (the solubilities of NbC_{1-x} and TiC_{1-x} in δ-$WC_{1\pm x}$ are very low) *See also* section C–Nb–Ti–W in Table I-2.14	
NbC_{1-x}–TiC_{1-x}–ZrC_{1-x}	–	1900–2000	Monocarbide (cubic) continuous solid solution (complete solubility in the system)	[5, 81, 477]
NbC_{1-x}–δ-$TiN_{1\pm x}$	–	~2600	Monocarbonitride continuous solid solution (complete solubility in the system)	[5, 495, 571, 572, 1030]
NbC_{1-x}–$TiO_{1\pm x}$	–	–	Terminal mutual solid solubilities between the components (?)	[5]
NbC_{1-x}–$UC_{1\pm x}$	–	1200–1800 1900–2100	Monocarbide extended solid solutions (?) Monocarbide continuous solid solution (a spacing-composition curve in the system is almost linear) *See also* section C–Nb–U in Table I-2.14	[5, 69, 81, 477, 484, 496–498, 727, 937]
NbC_{1-x}–$UC_{1\pm x}$–VC_{1-x}	–	1900–2050	Monocarbide continuous solid solution with great miscibility gap because of low mutual solubilities in the $UC_{1\pm x}$–VC_{1-x} system *See also* section C–Nb–U–V in Table I-2.14	[5, 69, 477]
NbC_{1-x}–$UC_{1\pm x}$–δ-$WC_{1\pm x}$	–	1900–2000	The solubility of δ-$WC_{1\pm x}$ in NbC_{1-x}–$UC_{1\pm x}$ monocarbide (cubic) continuous solid solution increases with increasing content of Nb from low values up to ~30–35 mol.%	[5]

(continued)

Table 4.19 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	Reference
NbC_{1-x}–$UC_{1\pm x}$–ZrC_{1-x}	–	1400–1800	Monocarbide (cubic) continuous solid solution with a miscibility gap because of lower mutual solubilities in the NbC_{1-x}–$UC_{1\pm x}$ system; the miscibility gap shortens noticeably with increasing temperature, mainly due to increasing the solubility of NbC_{1-x} in $UC_{1\pm x}$	[5, 69, 81, 477, 484, 727, 940–942, 947, 1099]
	–	1900–2050	Monocarbide (cubic) continuous solid solution (complete solubility in the system)	
	–	~2400	$(Nb,Zr,U)C_{1+x}$ hyper-stoichiometric mixed carbide phase based on $ZrC_{\sim 1.0}$ with 5–10 mol.% UC_{1+x} is formed from powdered individual carbides mixture (exposure >20 min)	
	–	>2530	$(Nb,Zr,U)C_{1-x}$ hypo-stoichiometric mixed carbide phase (x = 0.05) based on $ZrC_{\sim 1.0}$ with 5–10 mol.% U carbide is formed from the powdered mixture of Zr and Nb carbides with UH_3 and C addition *See also* section C–Nb–U–Zr in Table I-2.14	
β-$Nb_{2+x}C$–$UC_{1\pm x}$			*See* section C–Nb–U in Table I-2.14	
NbC_{1-x}–ζ-U_2C_3			*See* section C–Nb–U in Table I-2.14	
NbC_{1-x}–α-UC_{2-x}	–	1700	The components are compatible; no solubility of NbC_{1-x} in α-UC_{2-x} *See also* section C–Nb–U in Table I-2.14	[465, 498]
β-$Nb_{2+x}C$–α-UC_{2-x}			*See* section C–Nb–U in Table I-2.14	
NbC_{1-x}–UN_{1-x}	–	–	Monocarbonitride (cubic) continuous solid solution (complete solubility in the system?)	[5]
NbC_{1-x}–$UP_{1\pm x}$	–	–	Terminal mutual solid solubilities between the components (?)	[5]
NbC_{1-x}–$US_{1\pm x}$	–	–	Terminal mutual solid solubilities between the components (?)	[5]
NbC_{1-x}–VC_{1-x}	–	1300	Monocarbide extended solid solutions; the miscibility gap ranges from ~$Nb_{0.80}V_{0.20}C_{1-x}$ to ~$Nb_{0.15}V_{0.85}C_{1-x}$	[5, 69, 81, 145, 446, 476, 493, 494, 499, 817, 818, 892, 897, 899, 939, 1030, 1094]
	–	1500	Monocarbide extended solid solutions; the miscibility gap ranges from ~$Nb_{0.60}V_{0.40}C_{1-x}$ to ~$Nb_{0.22}V_{0.78}C_{1-x}$	
	–	From 1600–1700 (or ~1450, ?) up to 2700	Monocarbide (cubic) continuous solid solution (the variation of lattice parameter a, nm with composition for $(Nb_{1-y}V_y)C_{1-x}$ phases ($x \approx 0, 0 \leq y \leq 1$) is non-linear: $a = 0.4470 - 0.0216y - 0.0082y^2$); critical point of the miscibility gap is corresp. to ~1600–1700 °C (or 1400–1480 °C ?) and ~$Nb_{0.35–0.45}V_{0.55–0.65}C_{1-x}$ composition *See also* section C–Nb–V in Table I-2.14	

(continued)

Table 4.19 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	Reference
NbC_{1-x}–VC_{1-x}–δ-VN_{1-x}	–	–	Monocarbonitride (cubic) continuous solid solution (complete solubility in the system, the variation of lattice parameter a, nm with composition for $(Nb_{1-y}V_y)(C_{1-x}N_x)$ phases ($0 \leq x \leq 1, 0 \leq y \leq 1$): $a = 0.4470 - 0.0080x - 0.0298y + 0.0044xy$)	[34, 159, 899, 1030]
NbC_{1-x}–VC_{1-x}–δ-$WC_{1\pm x}$	–	–	Extended solid solution based on NbC_{1-x}–VC_{1-x} monocarbide continuous solid solution; the max. solubilities of δ-$WC_{1\pm x}$ in NbC_{1-x} and VC_{1-x} are ~30 and ~60 mol.%, respectively (the solubilities of NbC_{1-x} and VC_{1-x} in δ-$WC_{1\pm x}$ are very low)	[5, 477]
NbC_{1-x}–VC_{1-x}–ZrC_{1-x}	–	1900–2050	Monocarbide (cubic) continuous solid solution with the great miscibility gap limited by ~$(V_{0.99}Zr_{0.01})C_{1-x}$–~$(V_{0.03}Zr_{0.97})C_{1-x}$–~$(Nb_{0.50}V_{0.30}Zr_{0.20})C_{1-x}$ compositions	[5, 81, 477]
NbC_{1-x}–VC_{1-x}–δ-NbN_{1-x}–δ-VN_{1-x}	–	1000–1200	General consideration of the system	[494, 899]
NbC_{1-x}–δ-VN_{1-x}	–	–	Monocarbonitride (cubic) continuous solid solution (complete solubility in the system)	[5, 34, 159, 899, 1030]
β-$Nb_{2+x}C$–β-$V_{2\pm x}C$	–	1400–2200	Semicarbide (hexagonal) continuous solid solution	[145, 476, 499]
	–	2200–2450	Extended solid solution based on β-$Nb_{2+x}C$; the max. solubility of V in it decreases with temperature growth from ~48 to ~20 at.% *See also* section C–Nb–V in Table I-2.14	
γ-$Nb_{2\pm x}C$–β-$V_{2\pm x}C$	–	2450	Extended solid solution based on γ-$Nb_{2\pm x}C$; the max. solubility of V is ~1 at.%	[145, 499]
	–	2530–2650	Extended solid solution based on γ-$Nb_{2\pm x}C$; the max. solid solubility of V varies from 10 to 15 at.% *See also* section C–Nb–V in Table I-2.14	
NbC_{1-x}–β-W_2B_{5-x}	–	2250	Eutectic NbC_{1-x}–β-W_2B_{5-x}	[908]
NbC_{1-x}–γ-WC_{1-x}	–	~2550–2850	Monocarbide (cubic) continuous solid solution *See also* section C–Nb–W in Table I-2.14	[5, 69, 145, 440, 460, 461]
NbC_{1-x}–δ-$WC_{1\pm x}$	–	1700–2000	The max. solid solubility of W in NbC_{1-x} corresponds to ~$(Nb_{0.65-0.70}W_{0.30-0.35})C_{1-x}$ composition and that of Nb in δ-$WC_{1\pm x}$ is ~2 at.%; the solubility of W increases with increasing value of x index in NbC_{1-x}	[5, 81, 441, 460, 461, 837]
		2300–2500	Solubility of δ-$WC_{1\pm x}$ in NbC_{1-x} is <1 mol.%	
	–	~2500	The max. solubility of W in NbC_{1-x} corresponds to compos. ~$(Nb_{0.05-0.30}W_{0.70-0.95})C_{1-x}$ and that of Nb in δ-$WC_{1\pm x}$ is ~1.5 at.%	
	–	2690	The max. solid solubility of Nb in δ-$WC_{1\pm x}$ is ~2 at.%	
	–	2860	The max. solubility of W in NbC_{1-x} corresponds to compos. ~$(Nb_{0.25}W_{0.75})C_{1-x}$ *See also* section C–Nb–W in Table I-2.14	

(continued)

Table 4.19 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	Reference
NbC_{1-x}–δ-$WC_{1\pm x}$–ZrC_{1-x}	–	–	Extended solid solution based on NbC_{1-x}–ZrC_{1-x} monocarbide (cubic) continuous solid solution; the max. solubilities of δ-$WC_{1\pm x}$ in NbC_{1-x} and ZrC_{1-x} are ~30–35 mol.% (the solubilities of NbC_{1-x} and ZrC_{1-x} in δ-$WC_{1\pm x}$ are very low)	[5, 477]
NbC_{1-x}–α-$W_{2+x}C$	–	1700	The max. solid solubility of Nb in α-$W_{2+x}C$ is ~18 at.%; the effects of semicarbide phase transitions on phase equilibrium and solubility are not determined *See also* section C–Nb–W in Table I-2.14	[5, 441, 460, 461]
NbC_{1-x}–γ-$W_{2\pm x}C$	–	2490–2690	The max. solubility of Nb in γ-$W_{2\pm x}C$ varies from ~23 to ~25 at.% *See also* section C–Nb–W in Table I-2.14	[5, 441, 460, 461]
β-$Nb_{2+x}C$–α/β-$W_{2+x}C$	–	1700–2490	The max. solubility of W in β-$Nb_{2+x}C$ is ~1–2 the effects of semicarbide phase transitions on phase equilibrium and solubility are not determined *See also* section C–Nb–W in Table I-2.14	[441, 460, 461]
γ-$Nb_{2\pm x}C$–γ-$W_{2\pm x}C$	–	2690	The solubility of W in γ-$Nb_{2\pm x}C$ is very low; the effects of semicarbide phase transitions on phase equilibrium and solubility are not determined *See also* section C–Nb–W in Table I-2.14	[441, 460, 461]
NbC_{1-x}–$YC_{1\pm x}$	–	–	Terminal mutual solid solubilities between the components (?)	[5]
NbC_{1-x}–$YN_{1\pm x}$	–	–	Terminal mutual solid solubilities between the components (?)	[5]
NbC_{1-x}–$YbN_{1\pm x}$	–	–	Monocarbonitride (cubic) continuous solid solution (complete solubility in the system?)	[5]

(continued)

Table 4.19 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	Reference
NbC_{1-x}–ZrC_{1-x}	–	300	Monocarbide extended solid solutions; the miscibility gap ranges from $\sim Nb_{0.3}Zr_{0.7}C_{1-x}$ to $\sim Nb_{0.9}Zr_{0.1}C_{1-x}$ (theoretically calculated)	[5, 27, 69, 145, 209, 412, 441,
	–	~410–3400 (or 1600–3400, ?)	Monocarbide (cubic) continuous solid solution (the variation of lattice parameter a, nm with composition for $(Nb_{1-y}Zr_y)C_{1-x}$ phases ($x \approx 0, 0 \leq y \leq 1$) is linear: $a = 0.4470 + 0.0230y$); the critical point of the miscibility gap is corresponding to ~410–570 °C and $\sim Nb_{0.55–0.65}Zr_{0.35–0.45}C_{1-x}$ composition	463, 476, 482, 484, 491, 500–502, 522, 557, 558, 583, 759, 808, 817, 844, 887,
	–	>940	The homogeneity region limits of monocarbide (cubic) continuous solid solution phase are $ZrC_{0.60}$–$NbC_{0.70}$–$NbC_{\sim 1.0}$–$ZrC_{0.98}$ compositions	889, 892, 920, 939, 991, 1013, 1018, 1028,
	–	<1900	No solid solution is formed (1 h exposure) for a two carbides powdered mixture treated by hot pressing	1030, 1086, 1094, 1100]
	Ar, >0.1 MPa	1900–2000	Additions of NbC_{1-x} have no contribution to ZrC_{1-x} hot-pressing densification process, as the formation of solid solutions requires higher temperatures and longer time	
	–	–	Formation of $(Nb_{0.8}Zr_{0.2})_4C_3T_x$ (MXene, 2D-Nb_4C_3-based solid solution, where T_x is a surface termination) *See also* section C–Nb–Zr in Table I-2.14	
β-$Nb_{2+x}C$–ZrC_{1-x}	–	1500–1800	The solid solubility of Zr in β-$Nb_{2+x}C$ is low *See also* section C–Nb–Zr in Table I-2.14	[5, 69, 145, 441, 476]
NbC_{1-x}–ZrC_{1-x} –$ZrN_{1\pm x}$	–	–	Monocarbonitride (cubic) continuous solid solution (complete solubility in the system, the variation of lattice parameter a, nm with composition for $(Nb_{1-y}Zr_y)(C_{1-x}N_x)$ phases ($0 \leq x \leq 1, 0 \leq y \leq 1$): $a = 0.4470 - 0.0080x + 0.0230y - 0.0045xy$)	[407, 495, 842, 1030]
NbC_{1-x}–$ZrN_{1\pm x}$	–	2450	Monocarbonitride (cubic) continuous solid solution (complete solubility in the system)	[5, 30, 407, 495, 842]
NbC_{1-x}–$ZrP_{1\pm x}$	–	–	Terminal mutual solid solubilities between the components (?)	[5]

(continued)

Table 4.19 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	Reference
NbC_{1-x}–β/γ-ZrO_{2-x}	–	>1300	Formation of mixed (complex) oxides and carbides in the contact zone between the compact dense NbC_{1-x} and powdered ZrO_{2-x}	[1, 5, 480, 481, 503, 504, 560, 582, 737]
	Vacuum 0.1–1 Pa	~1400–2100	Mass loss in the powdered mixtures with a simultaneous decrease of the content of combined carbon in them and formation of mixed phases $Nb_{1-z}Zr_zC_{1-x}$ and $Zr_{1-v}Nb_vO_{2-y}$	
	Ar	1600–1650	No new phases are formed in NbC_{1-x} + β-ZrO_{2-x} powdered (2.3 and 0.8 μm mean grain sizes, respectively) mixtures treated by hot pressing (up to 1 h exposure)	
	Vacuum 10 Pa	≥2100	The formation of intermediate phases in the contact zone between the compact dense materials *See also* section C–Nb–O–Zr in Table I-2.14	

[a]Practically, due to the serious difficulties in manufacturing transition metal carbide materials non-contaminated in any degree by O and N, all the materials labeled in literature as NbC_{1-x} with the appropriate certification more likely would have to be considered as $NbC_{1-x}N_yO_z$ with low or very low values of y and z indexes

Table 4.20 Chemical interaction of niobium carbide phases with gaseous media at elevated, high and ultra-high temperatures (reaction systems are given in alphabetical order)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	Reference
NbC_{1-x}–CO	CO	–	Formation of oxycarbide phases $NbC_{1-x}O_y$ (extended substitution solid solution based on NbC_{1-x} with approximate homogeneity limits: ~$NbC_{0.70}$– ~$NbC_{0.50}O_{0.15}$ – ~$NbC_{0.35}O_{0.45}$– ~$NbC_{0.35}O_{0.65}$– ~$NbC_{0.99}$) *See also* section C–Nb–O in Table I-2.14	[59, 828]
NbC_{1-x}–CO_2	CO_2	–	Formation of oxycarbide phases $NbC_{1-x}O_y$ (extended substitution solid solution based on NbC_{1-x} with approximate homogeneity limits: ~$NbC_{0.70}$– ~$NbC_{0.50}O_{0.15}$– ~$NbC_{0.35}O_{0.45}$– ~$NbC_{0.35}O_{0.65}$– ~$NbC_{0.99}$) and subsequent formation of oxide scales *See also* section C–Nb–O in Table I-2.14	[59, 828]

(continued)

Table 4.20 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	Reference
NbC_{1-x}–Cl_2	Cl_2	400–800	The interaction leads to the volatilization of $NbCl_5$ and formation of surface porous graphite or carbonaceous "cinder" layer growing in the accordance with linear law (apparent activation energy E = 38 kJ mol^{-1}): $NbC_{1-x} + 2½Cl_2 = NbCl_5 \uparrow + (1 - x)C$; at higher temperatures the diffusion resistance of the layer is more considerable, so the reaction rate at 600 °C is higher than at 800 °C due to some changes in the carbide-derived carbon layer structure[a]	[5, 64, 510, 511, 896]
		700–900	Powdered NbC_{1-x} decomposes easily; remaining carbon presents the ordered structure, which is composed of concentric wavy graphene layers intermediate between onion-like carbon and carbon blacks (*see* Table I-2.3) with high surface area (up to ~1300 m^2 g^{-1})	
NbC_{1-x}–H_2	H_2, 0.1 MPa	25–1000	The solubility of H in NbC_{1-x} amounts to 5 at.% (?)	[5, 17, 44, 83, 177, 210, 249, 542–545, 574, 727, 805, 824–826, 835, 1037]
	H_2, 0.1 MPa	700–1000	Formation of cubic carbohydride phases NbC_xH_y ($0.64 \le x \le 0.92$, $0 \le y \le 0.47$, ?)	
	H_2, 0.1 MPa	730–1330	The mass loss of compact NbC_{1-x} (32% porosity) observed in the range of ~0.1–2% (1 h exposure) is a consequence of the formation and subsequent decomposition of the oxide phases due to residual O_2; in comparison with vacuum (<1.5 Pa) H_2 promotes the mass loss of NbC_{1-x} due to the formation of NbH_x (?) and its subsequent decomposition	
	H_2	2000–2500	Weak interaction with the minimally determined mass changes (?)	
	H_2 (0.07–0.1% O_2)	2180	Transformation of monocarbide phase into semicarbide after the exposure of 100 min; the gradient of C concentration is insignificant, though $Nb_{2+x}C$ grains are observed all over the cross-section	
	H_2 (~0.1% O_2)	2230	Active growth of $Nb_{2+x}C$ grains is observed at the surface layers of materials	
	H_2	2700	High resistance (if there is no gas flow or low gas exchange rate?)	
	H_2 (0.07–0.1% O_2)	2710–2730	Intensive decarburization with the formation of CH_4 and damage of bulk parts; materials loss – 2 × 10^{-2} g cm^{-2} (exposure – 100 min)[b]	
	H_2 (0.07–0.1% O_2)	2830	Decarburization causes inhomogeneous concentration of C, changes in structure, decrease in the density and reduction in the strength (exposure – 17 min) Data available in literature are very controversial *See also* section C–H–Nb in Table I-2.14	

(continued)

Table 4.20 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	Reference
α/β-$Nb_{2+x}C$–H_2	H_2, 0.1 MPa	25–1000	The solubility of H in α/β-$Nb_{2+x}C$ amounts to 11 at.% (?)	[83, 542–545, 855]
	H_2, 0.1–100 kPa	400–600	Equilibrium H_2 pressures for the solid solutions containing from 1 to 10 at.% H	
	H_2, 0.1 MPa	1000	Formation of hexagonal carbohydride phases NbC_xH_y ($x \approx 0.5$, $0 \leq y \leq 0.6$, ?) *See also* section C–H–Nb in Table I-2.14	
NbC_{1-x}–N_2	N_2, 0.1 MPa	1100–1400	Formation of $NbC_{0.97}N_{0.02-0.03}$ carbonitride phase (15–24 h exposure)	[1, 5, 69, 82, 159, 170, 177, 485–487, 533, 548]
	N_2, 0.1–30 MPa	1100–1800	Formation of various carbonitride NbC_xN_z phases	
	N_2	1250–1450	Formation of various carbonitride NbC_xN_z and δ-NbN_{1-x} nitride phases (15 h exposure)	
	N_2	2400–2500	No interaction	
	N_2	~3500	Noticeable decarbonization of carbide phases *See also* section C–N–Nb in Table I-2.14	
β-$Nb_{2+x}C$–N_2	N_2	1250–1450	Formation of various semicarbonitride $Nb_2C_xN_z$ phases *See also* section C–N–Nb in Table I-2.14	[69, 485]
NbC_{1-x}–NH_3	NH_3	1000–1200	The formation of carbonitride and nitride phases	[82, 514]
NbC_{1-x}–O_2[c, d]	O_2 (~0.8 Pa)[e]-Ar	20–800	The oxidation of small pieces (irregular shape, uniform thickness – 0.8 mm, mass – 0.1–1 mg) of single crystal $NbC_{0.96}$ (100) in gaseous mixture stream (4.5 ppm O_2) with heating rate ~10 K min^{-1} leads to the maximal reaction rate 1.35×10^{-10} mol s^{-1} at 425 °C with subsequent rate decline up to $\sim 0.4 \times 10^{-10}$ mol s^{-1} at 800 °C; the reactivity of (*hkl*) planes correlates with C occupancy in the unit cell of NbC_{1-x}	[5, 63, 67, 82, 93, 159, 177, 515–531, 547, 712, 717, 727, 743, 745, 746, 829–831]
	O_2 (~0.8 Pa)[e]-Ar	180–360	In this temperature range the oxidation of sintered $NbC_{\sim 1.0}$ in gas mixture flow (5 ppm O_2) with heating rate ~0.25 K min^{-1} leads to the maximal reaction rate 1.3×10^{-10} mol cm^{-2} s^{-1} at 250 °C with subsequent rate decline to $\sim 5 \times 10^{-12}$ mol cm^{-2} s^{-1} at >300 °C; after prolonged exposition (350 °C, 8 ppm O_2) the ratio of O/C = $\sim 1.5 \times 10^{-3}$ was achieved in solid	
	Air	350–550	Powdered nanocrystalline $NbC_{0.99}$ (average size of agglomerates ~60 nm) heating in the gas flow oxidizes with max. mass gain ~15% (maximum oxidation rate is at 480–550 °C)	

(continued)

Table 4.20 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	Reference
	Air	400–550	The oxidation of hot-pressed NbC_{1-x} (1–2% porosity) initiates by the formation of oxycarbide phases $NbC_{1-x}O_y$ (extended substitutional solid solution based on NbC_{1-x} with approximate homogeneity limits: $NbC_{0.7}$–$NbC_{0.5}O_{0.15}$–$NbC_{0.35}O_{0.45}$–$NbC_{0.35}O_{0.65}$–$NbC_{0.99}$) with the subsequent occurence of pentoxide α-$Nb_2O_5C_x$ (with dissolved carbon in it, ?) and separation of carbon[f] (sometimes metallic Nb and suboxides NbO_x were determined in the scale as well, ?); oxide scales have good adherence with substrate and oxidation kinetics is described by linear law at long-term stages (the oxidation resistance of NbC_{1-x} is higher than that of pure Nb)	
	O_2, 0.5–2.8 kPa	400–800	Isothermal/isobaric oxidation kinetics of bulk materials $NbC_{0.98}$ is described by linear-parabolic laws at the initial stages and linear law at the later stages, the effect of temperature is described by apparent activation energy $E \approx 50$–70 kJ mol^{-1} and effect of oxygen pressure – by power (exponent) index $m = 1$ with increasing at the later stages to $m = 1.5$	
	O_2/Ar, 4–16/ 36–24 kPa	420–600	Isothermal/isobaric oxidation kinetics of powdered $NbC_{0.92–0.96}$ (mean grain size – 0.6–3.4 μm, surface area – 0.97–2.2 m^2 g^{-1}) is described by Jander's equation and $E = 160 \pm 15$ kJ mol^{-1}; β-Nb_2O_{5-x} and NbO form oxide scale slightly cracked at the initial stages and largely fractured with further oxidation	
	Air	450	The oxidation mass gain of NbC_{1-x} is 1.4 mg cm^{-2} and 5.0 mg cm^{-2} for 1 h and 2 h exposure, respectively	
	O_2/Ar, 20/80 kPa	515–540	Powdered $NbC_{0.92–0.96}$ (mean grain size – 0.6–3.4 μm, surface area – 0.97–2.2 m^2 g^{-1}) heated from 350 °C with 5 °C min^{-1} rate performs instantaneous self-heating and combustion (sudden ~100% oxidation mass increase and CO_2 release marked with great/narrow exothermic DTA peak)	
	Air, flow	550–1000	Nanocrystalline $NbC_{0.99}$ preliminarily oxidized at 350–550 °C (relative degree of oxidation is ~60%) performs high gas corrosion resistance as its mass remains almost constant during further heating up to 1000 °C	

(continued)

Table 4.20 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	Reference
	O_2/Ar, 20/80 kPa	580–880	Single crystal $NbC_{0.95–0.97}$ (100) 0.05–2 mm in size heated with 5 °C min^{-1} rate performs gradual mass gain increase with increasing temperature (broad exothermic DTA peak with maximum around 800 °C)	
	Air	600	The oxidation mass gain of NbC_{1-x} is 11.7 mg cm^{-2} (1 h exposure)	
	Air	600–700	The oxidation of hot-pressed (1–2% porosity) NbC_{1-x} evolves by the formation of oxide scales containing α-$Nb_2O_5C_x$ (with dissolved carbon in it, ?), α-$NbO_{2\pm x}$ and dispersed carbon phase burning-off gradually (sometimes metallic Nb was determined in the scale as well, ?); oxide scales have no good adherence with substrate and oxidation kinetics is described by linear law at the initial stages and parabolic law at the later stages (the oxidation resistance of NbC_{1-x} is lower than that of pure metal Nb)	
	O_2, 5×10^{-4} Pa	600–700	The oxidation of synthetically grown NbC_{1-x} film (thickness 50 nm) gives the oxide scale (thickness 4.7–5.6 nm for 45 min exposure) containing γ-NbO_{2-x}, α-NbO_2, β-Nb_2O_{5-x} and α-Nb_2O_5 (the reactivity of NbC_{1-x} film is lower than that of pure metallic Nb film under the same conditions)	
	O_2, 0.03–13 kPa	600–900	The oxidation of sintered $NbC_{0.94}$ (1–5% porosity, mean grain size – 50–100 μm) leads to the formation of single phase α-Nb_2O_5 scale (the only traces of β-Nb_2O_{5-x} at 800 °C and 6.7 kPa, carbon burning-off in the scale is incomplete); oxidation kinetics is described by quasi-linear law with apparent activation energy $E = 92 \pm 8$ kJ mol^{-1} and power (exponent) index $m = 1.4 \pm 0.1$ (limiting stage is oxygen adsorption)	
	O_2/Ar, $0.8/\sim 10^5$ Pa	700–900	The effect of temperature on the oxygen consumption of sintered NbC_{1-x} (~40% porosity) is described by apparent activation energy $E = 96 \pm 5$ kJ mol^{-1}	
	Air	700–900	The oxidation rate of hot-pressed NbC_{1-x} (1–2% porosity) grows sharply, the oxide scale containing α-$NbO_{2\pm x}$, α-Nb_2O_5 and stabilized by carbon (?) β-Nb_2O_{5-x} loses its protective properties completely and crumbles away from the surface of samples; oxidation kinetics is described by linear law	

(continued)

Table 4.20 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	Reference
	Air flow, ~5 cm s^{-1}	800	Oxidation mass gain of hot-pressed NbC_{1-x} (porosity ~4%) is 57 mg cm^{-2} (0.5 h exposure)	
	O_2, 0.5-2.8 kPa	800–1100	The effect of temperature on the oxidation rate of sintered NbC_{1-x} is described by negative apparent activation energy and effect of oxygen pressure – by power (exponent) index $m = 1$ (0.5–1 kPa) and $m \approx 1/5$ (1–2.8 kPa)	
	O_2, 1.3 kPa	800–1100	The effect of temperature on the oxidation rate of arc-cast $NbC_{\sim 1.0}$ with 3 mas.% non-combined carbon is described by negative apparent activation energy	
	O_2, 0.03 kPa	900–1300	The oxidation mass gain of sintered $NbC_{0.94}$ (1–5% porosity, mean grain size – 50–100 μm) raises from 4.0 to 6.5 mg cm^{-2} (20 min exposure) with temperature growth	
	Air	1000–1200	The intensive (catastrophic) oxidation of hot-pressed NbC_{1-x} (1–2% porosity) accompanied with cracking and splitting of the oxide scales containing β-Nb_2O_{5-x} and α-$NbO_{2\pm x}$; porous samples are disintegrated completely	
	Air	1100–1400	Severe oxidation of materials	
	O_2, 0.03–13 kPa	1000–1400	The oxidation of sintered $NbC_{0.94}$ (1–5% porosity, mean grain size – 50–100 μm) leads to the formation of scale containing dense (more sintered) β-Nb_2O_{5-x} and α-$NbO_{2\pm x}$ (the only traces of α-Nb_2O_5 at 1000 °C, carbon content in the oxide scale is <0.2 mas.%, as an evidence of stoichiometric burning-off); oxidation kinetics is described by parabolic law with apparent activation energy $E \approx 60$–80 kJ mol^{-1} and power (exponent) index $m \approx ½$ (limiting stage is solid state diffusion through α-$NbO_{2\pm x}$ dense layer)	
	O_2, 0.05–13 kPa	1500–1600	The oxidation process of sintered $NbC_{0.94}$ (1–5% porosity) is described by linear law (limiting stage is oxygen adsorption on the outer surface of liquid oxide)	
	O_2, ~10^{-5} Pa	1700–1800	Treatment (annealing) of $NbC_{0.97}$ and $NbC_{0.74}$ powders (1 h exposure) leads to their severe decarburization and conversion of monocarbide phases into semicarbide β-$Nb_{2+x}C$ or metallic Nb phases, respectively	

(continued)

Table 4.20 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	Reference
	O_2, 0.1–0.001 Pa	1730–2230	During the oxidation of hot-pressed NbC_{1-x} a metallic surface layer (formed by CO release) simultaneously reacts with O_2 forming volatile metal oxides; steady-state layer thickness is reached, if the rates of layer growth and metal evaporation are equal in the ablation *See also* section C–Nb–O in Table I-2.14	
α/β-$Nb_{2+x}C$–O_2	Air	300	Formation of a new carbide phase with lower carbon content *See also* section C–Nb–O in Table I-2.14	[48, 82, 743]

[a]Similar to the ridge-effect phenomenon discovered by Shabalin [512, 513] in the oxidation process of carbide containing materials

[b]The total rate of materials loss V is maximal at the first moments of exposure; the effect of exposure time t, s and temperature T, K at 2550–3000 K is expressed by the following equation V (g cm^{-2} s^{-1}) = $0.058\exp(-0.15 \times 10^{-3}t)\exp[(-54{,}000 + 3.00 \times 10^{-3}t)/RT]$, where R is the gas constant [727]

[c]Fine niobium carbide powders are pyrophoric [514]

[d]For near-stoichiometric NbC_{1-x} the value of the Pilling-Bedworth ratio $\alpha = M_O d_C / M_C d_O = 2.22$, where M_O is molecular mass of the oxide phase formed on the oxidation of 1 mol of carbide phase, M_C is molecular mass of carbide phase, d_C and d_O are the densities of carbide and oxide phases, respectively [694]

[e]Partial pressure

[f]During O_2 consumption and oxide (scale) formation at temperatures <500–550 °C in NbC_{1-x} dense materials C does not oxidize

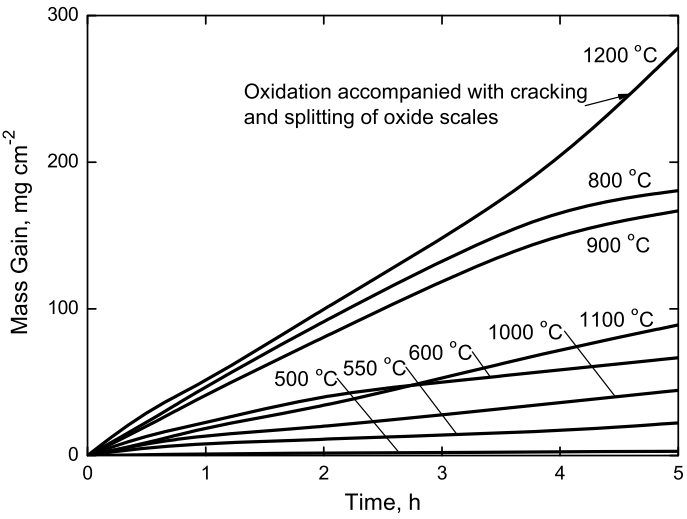

Fig. 4.37 Isothermal oxidation kinetics curves for 8 mm diameter disk-like samples of hot-pressed and subsequently annealed near-stoichiometric niobium monocarbide materials (porosity 1–2%) in air [518–520]

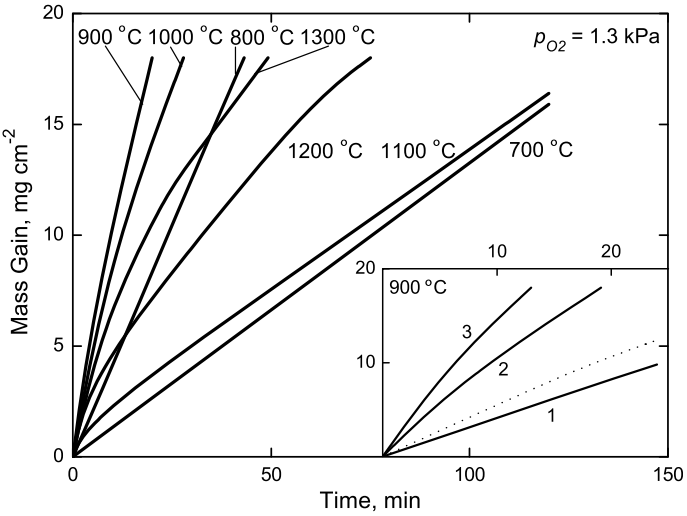

Fig. 4.38 Isothermal oxidation kinetics curves for flat-plate samples of sintered $NbC_{0.94}$ materials (porosity 1–5%) in oxygen gas flow at pressure of 1.33 kPa (*Inset* – carbon burning-off from the carbide phase: 1 – carbon loss curve (observed experimentally); 2 – thermogravimetric curve (measured directly); 3 – genuine (complete) mass gain due to the oxide formation (calculated); the dotted line corresponds to the stoichiometric oxidation of metal and carbon in the carbide phase, i.e. complete carbon burning-off in the oxide scale) [522, 523, 525]

Table 4.21 The parameters of wettability of niobium monocarbide phases with some liquid metals and alloys (melts)[a]

Melt (purity)	Atmo-sphere	Temp., °C	Time, s	$\gamma_{l\text{-}g}$, mJ m^{-2}	W_a, mJ m^{-2}	W_m^b, kJ mol^{-1}	θ, degree	Reference
				$NbC_{0.66}$				
Cu	Vacuum	1100	–	–	–	–	~70	[67, 540, 549]
				$NbC^c_{0.73}$				
Cu	Ar	1150	–	–	–	–	145 ± 3	[551]
				$NbC_{0.75}$				
Cu	Vacuum	1100	–	–	–	–	~90	[67, 540, 549]
				$NbC_{0.80}$				
Cu (99.99%)	Ar	1130	900	~1350	230	–	146 ± 3	[1, 534]
				$NbC^c_{0.81}$				
Cu	Ar	1150	–	–	–	–	146 ± 3	[551]

(continued)

Table 4.21 (continued)

Melt (purity)	Atmosphere	Temp., °C	Time, s	$\gamma_{l\text{-}g}$, mJ m^{-2}	W_a, mJ m^{-2}	W_m^b, kJ mol^{-1}	θ, degree	Reference
				$NbC_{0.82}$				
Cu	Vacuum	1100	–	–	–	–	~85	[67, 540, 549]
				$NbC^c_{0.91}$				
Cu	Ar	1150	–	–	–	–	150 ± 3	[551]
				$NbC_{0.92}$				
Cu (99.99%)	Ar	1130	900	~1350	315	–	140 ± 3	[1, 534]
				$NbC^d_{0.95}$				
Fe (C-4.0, Mn-1.4, Si-2.5, S-0.1%)	Pure Ar	1400–1450	900	–	–	–	~0	[1, 195, 541]
Fe (C-3.1, Mn-0.4, Si-1.6, S-0.02%)	Pure Ar	1400–1450	900	–	–	–	~0	[1, 195, 541]
Fe (C-2.7, Mn-0.4, Si-1.2, S-0.2%)	Pure Ar	1400–1450	900	–	–	–	~0	[1, 195, 541]
Fe (C-2.6, Mn-1.0, Si-1.0, Cr-0.3%)	Pure Ar	1400–1450	900	–	–	–	~0	[1, 195, 541]
Fe (C-1.0, Cr-1.4, Mn-0.4, Si-0.3%)	Pure Ar	1500–1550	900	–	–	–	~0	[1, 195, 541]
Fe (C-1.0, Cr-1.4, Mn-1.1, Si-0.6%)	Pure Ar	1500–1550	900	–	–	–	~0	[1, 195, 541]
Fe (C-0.8, Mn-0.2, Si-0.2, S-0.03%)	Pure Ar	1500–1550	900	–	–	–	~0	[1, 195, 541]
Fe (C-0.2, Cr-14.5, Ni-2.7, Mn-0.5%)	Pure Ar	1500–1550	900	–	–	–	~0	[1, 195, 541]
				$NbC_{0.97}$				
Cu (0.2% Ag)	Vacuum	1100	900	1250	1675	–	70	[1, 67, 535, 1080]
Cu (0.2% Ag)	Vacuum	1150	900	1255	1845	–	62	[1, 535]
Cu (0.2% Ag)	Vacuum	1200	900	1240	2070	–	48	[1, 67, 535, 549]
Cu (0.2% Ag)	Vacuum	1300	60	1360	2560	–	28	[1, 535]
Cu (0.2% Ag)	Vacuum	1300	600	1360	2655	–	18	[1, 535]
Cu (0.2% Ag)	Vacuum	1300	900	1360	2685	–	13	[1, 535]
Cu (0.2% Ag)	Vacuum	1300	1200	1360	2710	–	6	[1, 535]

(continued)

Table 4.21 (continued)

Melt (purity)	Atmosphere	Temp., °C	Time, s	$\gamma_{l\text{-}g}$, mJ m^{-2}	W_a, mJ m^{-2}	W_m^b, kJ mol^{-1}	θ, degree	Reference
				$NbC_{\sim 1.0}$				
Al	Vacuum	900	900	914	257	12.6	136 ± 0.3	[67, 195, 534, 536]
Al (99.97%)	Vacuum	920–1000	60–1200	914	155–166	–	145–146	[1, 534]
Al (99.97%)	Vacuum	1000	900	914	279	–	134 ± 3	[67, 195, 534, 549]
Al (99.97%)	Vacuum	1150	60–1200	914	155–355	–	127–146	[1, 534]
Au (99.999%)	–	~1065	1800	1140	1710	–	60	[1, 67, 540, 549]
Bi (99.999%)	Vacuum	320	900	390	114	5.9	135 ± 4	[1, 195, 534, 536]
Co (Si-0.3, Fe-0.3, Ni-0.2, Al-0.1%)	Vacuum	1420	300	1910	3760	–	11–14	[1, 67, 195, 535, 549]
Co (99.98%)	Vacuum, Ar	1450–1500	900	1805	>3610	–	~0	[1, 67, 195, 534, 536, 549]
Co[e]	Vacuum	1500	1200	1805	3665	–	11	[1, 537, 538]
Cu (99.99%)	Vacuum	1130–1300	900	1351	396	20.1	135	[1, 534, 536]
Cu	Vacuum	1100	–	–	–	–	~40	[67, 540]
Cu	Vacuum	1100	–	–	–	–	70	[67, 535]
Cu[f]	Vacuum, 1.3 mPa	1150	3600	2400	–	–	58	[1079]
Cu	Vacuum	1200	–	–	–	–	48	[67, 535]
Cu (Nb – 1.45%)[f]	Vacuum, 1.3 mPa	1150	3600	–	–	–	65	[1079]
Cu (Nb – 1.45, W – 2.8%)[f]	Vacuum, 1.3 mPa	1150	3600	–	–	–	56	[1079]
Fe (C-0.003, Si-0.3, Mg-0.3, Cu-0.2%)	Vacuum	1490	60	1900	3620	–	25	[1, 195, 535, 549]
Fe (99.999%)	Vacuum, Ar	1550	900	1780	>3560	–	~0	[1, 67, 195, 534, 536, 549]
Ga (99.99999%)	Vacuum	800	900	707	490	24.3	108 ± 0.3	[1, 195, 534, 536]
Ge (99.9999%)	Vacuum	1000	900	600	90	4.4	148 ± 2	[1, 195, 534, 536]
In (99.9995%)	Vacuum	250	900	559	70	3.4	151 ± 1	[1, 195, 260, 345]

(continued)

Table 4.21 (continued)

Melt (purity)	Atmo-sphere	Temp., °C	Time, s	$\gamma_{l\text{-}g}$, mJ m^{-2}	W_a, mJ m^{-2}	W_m^b, kJ mol^{-1}	θ, degree	Reference
Mn (99.8%)	Vacuum, Ar	1300	900	1750	>3500	–	~0	[1, 534, 536]
Ni (Fe-1.0, Al-0.3, Mg-0.2, Ag-0.2%)	Vacuum	1380	300	1810	3530	–	18	[1, 67, 195, 535]
Ni[g]	Vacuum	1400	900	1835	3550	–	21	[1, 539]
Ni	Vacuum	1450	–	–	–	–	21	[217, 832]
Ni (99.99%)	Vacuum, Ar	1450	900	1700	3400	–	~0	[1, 195, 534, 536]
Pb (99.98%)	Vacuum	400	900	480	73	3.6	148 ± 1	[1, 195, 534, 536]
Sb (99.999%)	Vacuum	700	900	384	246	12.1	111 ± 1	[1, 195, 534, 536]
Si (99.9999%)	Vacuum	1500	900	860	>1720	–	~0	[1, 195, 534, 536]
Sn (99.999%)	Vacuum	300	900	554	198	9.6	130 ± 1	[1, 195, 534, 536]
Tl (99.999%)	Vacuum	400	900	490	237	11.7	121 ± 1	[1, 195, 534, 536]

[a]The parameters of wettability are given in accordance with Young-Dupré equation $W_a = \gamma_{l\text{-}g} \times (1 + \cos\theta)$ and Young's equation $\gamma_{s\text{-}l} = \gamma_{s\text{-}g} - \gamma_{l\text{-}g}\cos\theta$, where W_a is the work of adhesion, $\gamma_{l\text{-}g}$ is the liquid-vapour interfacial energy (surface tension), $\gamma_{s\text{-}l}$ is solid-liquid interfacial energy, $\gamma_{s\text{-}g}$ is the solid-vapour interfacial energy and θ is the wetting contact angle [1]; compositions of melts are given in mass (weight) percentage

[b]$W_m = W_a(M/d)^{2/3}N_A^{1/3}$, where W_m is the molar work of adhesion, M is the molecular mass and d is the density of chemical compound, N_A is the Avogadro constant [1060]

[c]Porosity ≤15%

[d]Sintered materials (porosity – 12–16%)

[e]$\gamma_{s\text{-}l}$ = 480 mJ m^{-2}

[f]Sintered materials (porosity – ≤2%, contents: non-combined C – 0.40%, O – 0.01%, N – 0.26%, Ti – 0.03%)

[f]$\gamma_{s\text{-}l}$ = 1290 ± 390 mJ m^{-2}

Shabalin [512, 513] with ridge temperatures nearby 800–900 °C. Data on the catalytic activities of niobium carbides are given in several works [1, 483, 590–603, 820–822, 865, 866, 870, 976, 1010]; the electrochemical behaviour of niobium carbides in various media such as salt melts and aqueous solutions is described in [603–606, 688, 708, 823]. The parameters of wettability of niobium monocarbide phases with some liquid metals (melts) are listed in Table 4.21, the diffusion rates for the systems containing niobium carbide phases are presented in species pairs within the various ranges of temperatures in the Table 4.22. The characters of chemical interaction of niobium monocarbide with some common chemicals (acids, alkalies and salts in aqueous solutions) are summarized in Table 4.23.

Table 4.22 Diffusion rates and related parameters in the systems containing niobium, carbon and niobium carbides phases at various temperatures[a]

Species pair	Temperature dependence of the diffusion coefficient (diffusivity) $D = D_0\exp[(-E_A/R)/T]$, $cm^2\ s^{-1}$	Temperature range, K (°C)	Remarks on materials characteristics and measurement method	Reference
C → Nb	$10^{-2}\exp(-17{,}100/T)$	400–2610 (130–2340)	Combined data from several sources	[607, 608, 615, 616, 662]
	$1.09 \times 10^{-5}\exp(-16{,}000/T)$	1170–1470 (900–1200)	^{14}C radiometric method, sectioning technique	[5, 10, 11, 618, 623]
	$9.32 \times 10^{-3}\exp(-17{,}500/T)$	1370–1870 (1100–1600)	^{14}C radiometric method, sectioning technique	[5, 617, 622]
	$3.3 \times 10^{-3}\exp(-19{,}100)/T$	1200–2070 (930–1800)	Radiotracer ^{14}C with mechanical sectioning	[10, 11, 619]
	$2.6 \times 10^{-2}\exp(-19{,}000/T)$	2190–2610 (1920–2340)	Radiotracer ^{14}C with mechanical sectioning	[10, 11, 608, 621]
	$3.61 \times 10^{-2}\exp(-32{,}700/T)$	1370–1770 (1100–1500)	Radiometric method based on the formation of NbC_{1-x} layer	[1]
	$1.5 \times 10^{-2}\exp(-13{,}600/T)$	320–430 (50–160)	From the measurements of internal friction with the torsion pendulum	[611, 612]
	$4.5 \times 10^{-3}\exp[(-13{,}700 \pm 40)/T]$	330–430 (60–160)	From the measurements of internal friction	[5, 614]
	$(4.0 \pm 0.7) \times 10^{-3}\exp[(-16{,}600 \pm 90)/T]$	400–550 (130–280)	From the measurements of internal friction	[10, 11, 614, 623]
	$[(3.8–5.0) \pm (0.5–5.0)] \times 10^{-3} \times$ $\times \exp\{[-(16{,}600–16{,}800) \pm (50–500)]/T\}$	410–530 (140–260)	Determined by the various methods of internal friction measurements	[611, 613, 614]
	$1.94 \times 10^{-6}\exp(-8{,}300/T)$, ?	1270–2270 (1200–2000)	Calculation based on the thickness of NbC_{1-x} layer	[5, 63, 609]

(continued)

Table 4.22 (continued)

Species pair	Temperature dependence of the diffusion coefficient (diffusivity) $D = D_0\exp[(-E_A/R)/T]$, $cm^2\ s^{-1}$	Temperature range, K (°C)	Remarks on materials characteristics and measurement method	Reference
	$0.183\ \exp(-36{,}100/T)$	1570–1870 (1300–1600)	Calculation based on the growth of NbC_{1-x} phase layer	[1, 67, 620]
	$3.14 \times 10^4\exp[(-9{,}500 \pm 2{,}900)/T]$, ?	1270–2270 (1000–2000)	Calculation based on the growth of β-$Nb_{2+x}C$ phase layer	[609–611]
	$9.6 \times 10^{-2}\exp(-35{,}400/T)$	1570–1870 (1300–1600)	Calculation based on the growth of β-$Nb_{2+x}C$ phase layer	[1, 67, 620]
	$(0.45^{+0.13}_{-0.10})\exp[(-39{,}500 \pm 1{,}200)/T]$	1770–2370 (1500–2100)	Reaction diffusion coefficients, forward finite differences (FFD) method	[663]
$C \rightarrow NbC_{1-x}$	$0.86\exp(-44{,}000/T)$	2070–2470 (1800–2200)	Polycrystalline NbC_{1-x} ($x = 0.05$), ^{14}C radiotracer method	[2, 624]
	$8.8 \times 10^{-6}\exp(-17{,}800/T)$	1870–2270 (1600–2000)	Polycrystalline sintered NbC_{1-x} ($x = 0.25$, ~10–15% porosity), ^{14}C radiotracer	[1, 10, 11, 644, 695]
	$10^{-6}\exp(-16{,}300/T)$	1870–2270 (1600–2000)	Polycrystalline sintered NbC_{1-x} ($x = 0.02$, ~10–15% porosity), ^{14}C radiotracer	[10, 11, 644, 695]
	$(0.11^{+0.003}_{-0.1})\exp[(-47{,}300 \pm 900)/T]$	2670–3070 (2400–2800)	Polycrystalline (hot-pressed and annealed) NbC_{1-x} ($x \approx 0.03$, $Nb(C_{0.95}N_{0.04}O_{0.01})_{0.97}$, porosity <5%), ^{14}C radiotracer method	[2, 3, 625, 646, 727]
	6.62×10^{-10} 5.25×10^{-9} 3.24×10^{-8}	2500 (2230) 2800 (2530) 3100 (2830)	Calculated on the basis of developed statistic and thermodynamic theory of interstitial phases for NbC_{1-x} ($x = 0.03$)	[727]
	$(0.19^{+0.18}_{-0.16})\exp[(-48{,}100 \pm 1800)/T]$	2670–3070 (2400–2800)	Polycrystalline (hot-pressed and annealed) NbC_{1-x} ($x \approx 0.085$, $Nb(C_{0.998}N_{0.001}O_{0.001})_{0.915}$, porosity <5%), ^{14}C radiotracer method	[10, 11, 625, 656]
	$(0.21^{+0.18}_{-0.10})\exp[(-48{,}600 \pm 1800)/T]$	2670–3070 (2400–2800)	Polycrystalline (hot-pressed and annealed) NbC_{1-x} ($x \approx 0.085$, $Nb(C_{0.965}N_{0.032}O_{0.003})_{0.915}$, porosity <5%), ^{14}C radiotracer method	[2, 625, 646, 656, 945]
	$(1.47^{+0.52}_{-0.38})\exp[(-50{,}400 \pm 900)/T]$	2670–3070 (2400–2800)	Polycrystalline (hot-pressed and annealed) NbC_{1-x} ($x \approx 0.22$, $Nb(C_{0.94}N_{0.056}O_{0.004})_{0.78}$, porosity <5%), ^{14}C radiotracer method	[2, 625, 646, 656, 945]

(continued)

Table 4.22 (continued)

Species pair	Temperature dependence of the diffusion coefficient (diffusivity) $D = D_0\exp[(-E_A/R)/T]$, $cm^2\ s^{-1}$	Temperature range, K (°C)	Remarks on materials characteristics and measurement method	Reference
	2.61×10^{-9} 2.27×10^{-8} 1.31×10^{-7}	2500 (2230) 2800 (2530) 3100 (2830)	Calculated on the basis of developed statistic and thermodynamic theory of interstitial phases for NbC_{1-x} ($x = 0.21$)	[727]
	$0.5\exp[(-51{,}400 \pm 5{,}100)/T]$	2570–3070 (2300–2800)	Polycrystalline (hot-pressed and annealed) NbC_{1-x} ($x \approx 0.03$, $Nb(C_{0.95}N_{0.04}O_{0.01})_{0.97}$, 250–270 μm grain size, porosity <5%), ^{14}C radiotracer method	[2, 626, 637, 656, 945]
	$1.49 \times 10^{-2}\exp(-38{,}600/T)$	1900–3100 (1630–2830)	Polycrystalline NbC_{1-x} ($x \approx 0$, porosity $\leq 5\%$, $O + N \leq 0.32\%$), ^{14}C radiotracer method	[2, 627]
	$(2.59^{+1.82}_{-1.07})\exp[(-50{,}600 \pm 1{,}100)/T]$	1900–2315 (1630–2045)	Single crystal NbC_{1-x} ($x \approx 0.13$), plane (110), ^{14}C radiotracer method[b]	[2, 628, 629, 646, 661, 664]
	$(7.44^{+9.36}_{-4.14})\exp[(-52{,}900 \pm 1{,}700)/T]$	1900–2315 (1630–2045)	Single crystal NbC_{1-x} ($x \approx 0.17$), plane (100), ^{14}C radiotracer method[b]	[2, 628, 629, 646]
	$(2.22^{+1.98}_{-1.04}) \times 10^{-2}\exp[(-38{,}300 \pm 1{,}400)/T]$	1900–2315 (1630–2045)	Single crystal NbC_{1-x} ($x \approx 0.23$), plane (110), ^{14}C radiotracer method[b]	[2, 628, 629, 646, 664, 869]
	$7.6\exp(-44{,}600/T)$	1970–2370 (1700–2100)	Calculation based on the experimental investigations of solid-state NbC_{1-x}–Nb diffusion welding processes	[1, 636]
	$7.6\exp(-44{,}400/T)$	1970–2570 (1700–2300)	Calculation based on the experimental investigations of the reaction between metallic Nb and C (graphite)	[639, 647, 666, 695]
	$0.35\exp[(-37{,}300 \pm 1{,}400)/T]$	1970–2470 (1700–2200)	Calculation based on the growth of NbC_{1-x} phase layer during the reaction between metallic Nb and C (graphite)	[1, 67, 634, 646]
	$17.6\exp(-48{,}400/T)$	1970–2570 (1700–2300)	Parameters of temperature variation of NbC_{1-x} layer growth rate constant upon contact saturation of solid metallic Nb by C	[10, 11, 639]
	$1.08\exp(-41{,}800/T)$	2170–2570 (1900–2300)	Parameters of temperature variation of NbC_{1-x} layer growth rate constant upon contact saturation of solid metallic Nb by C	[10, 11, 640]
	$(0.265^{+0.29}_{-0.24})\exp[(-37{,}600 \pm 1{,}400)/T]$	1670–1970 (1400–1700)	Parameters of temperature variation of NbC_{1-x} layer growth rate constant upon contact saturation of solid metallic Nb by C	[665]

(continued)

Table 4.22 (continued)

Species pair	Temperature dependence of the diffusion coefficient (diffusivity) $D = D_0\exp[(-E_A/R)/T]$, $cm^2\ s^{-1}$	Temperature range, K (°C)	Remarks on materials characteristics and measurement method	Reference
	$7.8\exp(-42{,}300/T)$	1520–2150 (1250–1880)	Parameters of temperature variation of NbC_{1-x} layer growth rate constant upon surface carburization of Nb by pure C_3H_8	[672]
	$(4.5^{+12}_{-1.5}) \times 10^{-5}\exp[(-19{,}700 \pm 600)/T]$	1070–1770 (800–1500)	Parameters of temperature variation of NbC_{1-x} layer growth rate constant upon contact saturation of solid metallic Nb by C (in thin layers)[c]	[667–670]
	$6.84 \times 10^{-7}\exp(-11{,}000/T)$	1070–1270 (800–1000)	Parameters of temperature variation of NbC_{1-x} + β-$Nb_{2+x}C$ layer growth rate constant upon surface treatment of AISI 1040 steels in Nb containing powder mixtures (in thin layers)[c]	[675]
	$3.61 \times 10^{-2}\exp(-32{,}500/T)$	1370–1770 (1100–1500)	Parameters of reaction-chemical diffusion, contact saturation of solid Nb in mixture of C (graphite) and $BaCO_3$ rich in ^{14}C, radiometric method	[10, 11, 641]
	$0.102\exp[(-36{,}600 \pm 1{,}500)/T]$	1700–2100 (1425–1825)	Parameters of reaction-chemical diffusion, contact saturation of solid Nb (1 wt.% Zr) with C (graphite); metallography method	[10, 11, 642, 695]
	2.1×10^{-11}	1700 (1430)	Parameters of reaction-chemical diffusion, contact saturation of solid Nb (1 wt.% Zr) with C (graphite); metallography method	[642]
	$\sim 8.0 \times 10^{-7}$	2940 (2670)	Evaluation made from vaporization behaviour investigations	[160, 647, 695]
	$\sim\exp(-38{,}000/T)$	2770–2970 (2500–2700)	Initial composition $NbC_{0.64}$; mass loss method, evaluation made from vaporization behaviour investigations	[165, 166, 647]
	$\sim\exp[(-34{,}100 \pm 300)/T]$	1670–1970 (1400–1700)	Parameters of reaction-chemical diffusion, contact saturation of solid metallic Nb with C; metallography method[d]	[676]
	$1.2\exp(-39{,}800/T)$	2170–2570 (1900–2300)	Parameters of reaction-chemical diffusion, contact saturation of solid metallic Nb with C; metallography method	[10, 11, 640, 646, 647, 695]
	$2.65 \times 10^{-2}\exp(-32{,}700/T)$	2470–2810 (2200–2540)	Parameters of reaction-chemical diffusion during the formation of NbC_{1-x} ($x = 0.13$) obtained by precipitation from gas phase onto surface of graphite; X-ray method	[10, 11, 643]

(continued)

Table 4.22 (continued)

Species pair	Temperature dependence of the diffusion coefficient (diffusivity) $D = D_0\exp[(-E_A/R)/T]$, $cm^2\ s^{-1}$	Temperature range, K (°C)	Remarks on materials characteristics and measurement method	Reference
	$5.9\exp(-46{,}700/T)$	2420–3290 (2150–3015)	Parameters of chemical diffusion, decarburization of NbC_{1-x} materials	[10, 11, 645, 647]
	$0.66\exp(-37{,}000/T)$	2170–2800 (1900–2530)	Chemical diffusion coefficients calculated on the basis of experiments on deposition of non-stoichiometric carbide NbC_{1-x} ($0.10 \leq x \leq 0.36$, porosity $\leq$ 10%) from $CH_4 + H_2$ gas phase[e]	[1, 67, 635]
	$16.6\exp(-48{,}800/T)$	2570–3570 (2300–3300)	Averaged calculation for NbC_{1-x} in homogeneity range ($0.025 \leq x \leq 0.36$)	[635, 657, 658]
	$3.84 \times 10^{-5}\exp(23.24x)\exp[(-28{,}450 \pm 850)/T]$	1670–1970 (1400–1700)	Interdiffusion coefficients as a function of composition, bulk diffusion couples method based on electron probe microanalysis	[665]
	$10^{(3.6-9.8x)}\exp\{-[(3.6 - 5.6x) \times 23{,}000]/T\}$	2570–3570 (2300–3300)	Averaged calculation for NbC_{1-x} ($0.025 \leq x \leq 0.36$) on the basis of steady state creep behaviour data	[635, 657, 658, 666]
	$(0.4^{+0.4}_{-0.2})\exp[(-44{,}800 \pm 1{,}500)/T]$	1770–2370 (1500–2100)	Reaction diffusion coefficients, profile fitting method (high concentrations)	[663]
	$(0.12^{+0.4}_{-0.2})\exp[(-38{,}900 \pm 2{,}300)/T]$	1770–2370 (1500–2100)	Reaction diffusion coefficients, profile fitting method (low concentrations)	[663]
	$(2.3^{+3.0}_{-1.3})\exp[(-46{,}100 \pm 1{,}600)/T]$	1770–2370 (1500–2100)	Reaction diffusion coefficients, forward finite differences (FFD) method	[663]
	$6.5 \times 10^{-2}\exp(-41{,}000/T)$	–	The values of chemical diffusion coefficients summarized on the basis of several works	[178]
	$\sim\exp[(-41{,}600 \pm 600)/T]$	2000–2500 (1730–2230)	Parameters of reaction-chemical diffusion measured by diffusion couples method	[839]
	$\sim 5 \times 10^{-8}$	3170 (2900)	Polycrystalline (hot-pressed and annealed) NbC_{1-x} ($x = 0.04$), ^{14}C radiotracer method	[637, 834]
	–	2470–2570 (2200–2300)	Non-isothermal parameters of temperature variation of NbC_{1-x} layer growth rate constant upon surface carburization of Nb by pure CH_4; mass gain, layer thickness growth and heat release rate methods[f]	[677]

(continued)

Table 4.22 (continued)

Species pair	Temperature dependence of the diffusion coefficient (diffusivity) $D = D_0\exp[(-E_A/R)/T]$, $cm^2\ s^{-1}$	Temperature range, K (°C)	Remarks on materials characteristics and measurement method	Reference
C → β-$Nb_{2+x}C$	$5.7 \times 10^6\exp(-15{,}000/T)$	1870–2270 (1600–2000)	Polycrystalline sintered β-$Nb_{2+x}C$ (~17% porosity), ^{14}C radiotracer method	[10, 11, 644]
	$2.3 \times 10^{-3}\exp[(-20{,}100 \pm 2000)/T]$	1970–2470 (1700–2200)	Calculation based on the growth of β-$Nb_{2+x}C$ phase layer during the reaction between metallic Nb and C (graphite)	[1, 67, 634, 646, 666]
	$0.3\exp(-40{,}600/T)$	1970–2570 (1700–2300)	Parameters of temperature variation of β-$Nb_{2+x}C$ layer growth rate constant upon contact saturation of solid metallic Nb by C	[10, 11, 639]
	$40.2\exp(-30{,}700/T)$	2170–2570 (1900–2300)	Parameters of temperature variation of β-$Nb_{2+x}C$ layer growth rate constant upon contact saturation of solid metallic Nb by C	[10, 11, 640]
	$(0.157^{+0.19}_{-0.13})\exp[(-36{,}400 \pm 3{,}100)/T]$	1670–1970 (1400–1700)	Parameters of temperature variation of β-$Nb_{2+x}C$ layer growth rate constant upon contact saturation of solid metallic Nb by C	[665]
	$0.10\exp(-36{,}100/T)$	1520–2150 (1250–1880)	Parameters of temperature variation of β-$Nb_{2+x}C$ layer growth rate constant upon surface carburization of Nb by pure C_3H_8	[672]
	$(3.7^{+10}_{-2}) \times 10^{-4}\exp[(-22{,}900 \pm 1800)/T]$	1070–1470 (800–1200)	Parameters of temperature variation of β-$Nb_{2+x}C$ layer growth rate constant upon contact saturation of solid metallic Nb by C (in thin layers)[c]	[667–670]
	$\sim\exp[(-34{,}600 \pm 1{,}500)/T]$	1700–2100 (1425–1825)	Parameters of reaction-chemical diffusion, contact saturation of solid Nb (1 wt.% Zr) with C (graphite); metallography method	[10, 11, 642]
	$\sim\exp(-40{,}600/T)$	1970–2570 (1700–2300)	Parameters of temperature variation of β-$Nb_{2+x}C$ layer growth rate constant upon contact saturation of solid metallic Nb by C	[10, 11, 639]
	$\sim\exp[(-36{,}000 \pm 500)/T]$	1670–1970 (1400–1700)	Parameters of reaction-chemical diffusion, contact saturation of solid metallic Nb with C; metallography method[d]	[676]
	$(2.9^{+2.3}_{-1.3})\exp[(-45{,}800 \pm 1{,}200)/T]$	1770–2370 (1500–2100)	Reaction diffusion coefficients, forward finite differences (FFD) method	[178, 663]
	$(2.04 \pm 0.57)\exp[(-35{,}500 \pm 2{,}200)/T]$	1670–1970 (1400–1700)	Interdiffusion coefficients, bulk diffusion couples method based on electron probe microanalysis	[665]
	$\sim\exp[(-35{,}500 \pm 600)/T]$	2000–2500 (1730–2230)	Parameters of reaction-chemical diffusion measured by diffusion couples method	[839]

(continued)

Table 4.22 (continued)

Species pair	Temperature dependence of the diffusion coefficient (diffusivity) $D = D_0\exp[(-E_A/R)/T]$, $cm^2\ s^{-1}$	Temperature range, K (°C)	Remarks on materials characteristics and measurement method	Reference
	–	2470–2570 (2200–2300)	Non-isothermal parameters of temperature variation of β-$Nb_{2+x}C$ layer growth rate constant upon surface carburization of Nb by CH_4; mass gain, layer thickness growth and heat release rate methods[f]	[677]
C → $(Nb_{1-y}Zr_y)C_{1-x}$	$(2.28^{+0.47}_{-0.30})\exp[(-48{,}600 \pm 1{,}300)/T]$	2770–3470 (2500–3200)	Polycrystalline $(Nb_{1-y}Zr_y)C_{1-x}$ ($x = 0.03$, $y = 0.48$), ^{14}C radiotracer method	[2, 727]
	8.26×10^{-9} 6.72×10^{-8} 3.56×10^{-7}	2500 (2230) 2800 (2530) 3100 (2830)	Calculated on the basis of developed statistic and thermodynamic theory of 3-component interstitial phases for $(Nb_{1-y}Zr_y)C_{1-x}$ ($x = 0.10$, $y = 0.48$)	[727]
	$(0.84^{+0.47}_{-0.30})\exp[(-50{,}600 \pm 1{,}700)/T]$	2770–3470 (2500–3200)	Polycrystalline $(Nb_{1-y}Zr_y)C_{1-x}$ ($x = 0.18$, $y = 0.48$), ^{14}C radiotracer method	[2, 727]
	1.36×10^{-9} 1.20×10^{-8} 6.91×10^{-8}	2500 (2230) 2800 (2530) 3100 (2830)	Calculated on the basis of developed statistic and thermodynamic theory of 3-component interstitial phases for $(Nb_{1-y}Zr_y)C_{1-x}$ ($x = 0.20$, $y = 0.48$)	[727]
Nb → NbC_{1-x}	$(0.11^{+0.28}_{-0.08})\exp[(-64{,}000 \pm 3{,}500)/T]$	2670–3070 (2400–2800)	Polycrystalline hot-pressed NbC_{1-x} ($0.03 \le x \le 0.22$, porosity <5%), ^{95}Nb radiotracer[g]	[2, 3, 10, 11, 625, 646, 727]
	8.73×10^{-13} 1.32×10^{-11} 1.23×10^{-10}	2500 (2230) 2800 (2530) 3100 (2830)	Calculated on the basis of developed statistic and thermodynamic theory of interstitial phases for NbC_{1-x} ($0.03 \le x \le 0.21$)	[727]
	$0.7\exp(-64{,}400/T)$	2770–3270 (2500–3000)	Polycrystalline hot-pressed NbC_{1-x} ($x \approx 0.03$, $Nb(C_{0.95}N_{0.04}O_{0.01})_{0.97}$, porosity <5%), ^{95}Nb radiotracer method	[10, 11, 637]
	$1.7\exp(-57{,}900/T)$	2670–3170 (2400–2900)	Polycrystalline hot-pressed NbC_{1-x} ($x \approx 0.18$, porosity <5%), ^{95}Nb radiotracer method	[10, 11, 637]
	$(4.54^{+2.85}_{-1.75})\exp[(-70{,}500 \pm 1{,}200)/T]$	2370–2660 (2100–2390)	Single crystal NbC_{1-x}, ($0.13 \le x \le 0.23$), ^{95}Nb radiotracer method[g]	[2, 630, 631, 646, 661, 664]

(continued)

Table 4.22 (continued)

Species pair	Temperature dependence of the diffusion coefficient (diffusivity) $D = D_0\exp[(-E_A/R)/T]$, $cm^2\ s^{-1}$	Temperature range, K (°C)	Remarks on materials characteristics and measurement method	Reference
	$5.5 \times 10^{-6}\exp(-27{,}700/T)$	1970–2370 (1700–2100)	Polycrystalline sintered NbC_{1-x} ($0.02 \le x \le 0.25$, ~17% porosity), ^{95}Nb radiotracer method[g]	[1, 67, 644, 695]
	$\sim 10^{-9}$	3170 (2900)	Polycrystalline (hot-pressed and annealed) NbC_{1-x} ($x = 0.04$), ^{14}C radiotracer method	[637, 834]
	$1.25\exp(-54{,}300/T)$	–	Values calculated on the theoretical basis	[1, 67, 638]
	$\sim\exp(-67{,}000/T)$	–	Estimated on the basis of LeClaire's relation (connected with phase melting point)	[695]
Nb → $(Nb_{1-y}Zr_y)C_{1-x}$	$(0.51^{+1.33}_{-0.50}) \times 10^2\exp[(-77{,}100 \pm 4900)/T]$	2770–3470 (2500–3200)	Polycrystalline $(Nb_{1-y}Zr_y)C_{1-x}$ ($0.03 \le x \le 0.18$, $y = 0.48$), ^{95}Nb radiotracer method[g]	[2, 727]
	2.12×10^{-12} 5.71×10^{-11} 8.26×10^{-10}	2500 (2230) 2800 (2530) 3100 (2830)	Calculated on the basis of developed statistic and thermodynamic theory of 3-component interstitial phases for $(Nb_{1-y}Zr_y)C_{1-x}$ ($0.10 \le x \le 0.20$, $y = 0.48$)	[727]
Nb → $(Nb_{0.5}Ti_{0.5})C_{1-x}$	$4.7 \times 10^2\exp(-60{,}400/T)$	–	Sintered $(Nb_{0.5}Ti_{0.5})C_{1-x}$ ($x = 0.15$), ^{95}Nb radiotracer method	[654, 655, 695]
Pm → NbC_{1-x}	$(1.08^{+0.327}_{-0.98}) \times 10^6\exp[(-61{,}400 \pm 4{,}500)/T]$	2270–2820 (2000–2550)	Polycrystalline hot-pressed NbC_{1-x} ($x \approx 0.03$–0.22), ^{147}Pm radiotracer method[g]	[2, 633]

[a]The chemical diffusion coefficient concentration and temperature dependencies can be represented by an equation of the form $D' = D'_0\exp(\beta x)\exp[(-E'_A/R)/T]$, where β is the concentration factor and x is the value of index in NbC_{1-x} formula (for the ranges of $x = 0.01$–0.60 and $T = 1730$–3030 °C the recommended values are $D'_0 = 6.26 \times 10^{-2}\ cm^2\ s^{-1}$, $\beta = 10$ and $E'_A = 326\ kJ\ mol^{-1}$ [647]), and the discrepancy between chemical and self-diffusivities may be estimated on the basis of the thermodynamical relationship: $D' = D \times [1 + (d\ln\gamma_C/d\ln N_C)]$, where γ_C is the activity coefficient of carbon and N_C is the atomic fraction of carbon (evaluated at 2000 °C in NbC_{1-x}: at $x = 0.01$, $d\ln\gamma_C/d\ln N_C = 141$ and at $x = 0.20$, $d\ln\gamma_C/d\ln N_C = 15$ [652]); the approximate values of apparent activation energy for some diffusion controlled processes in NbC_{1-x} and β-$Nb_{2+x}C$: (**a**) collective recrystallization of loosely-poured powders – 160 kJ mol^{-1} ($NbC_{0.77}$, 1600–2000 °C) [653], 175 kJ mol^{-1} ($NbC_{0.82}$, 1600–2000 °C) [653], 195 kJ mol^{-1} ($NbC_{0.95}$, 1600–2000 °C) [653], 245 kJ mol^{-1} ($NbC_{0.99}$, 1600–2000 °C) [653]; (**b**) recrystallization (grain growth) – 300 kJ mol^{-1} (sintered after bulk deformation $NbC_{\sim 1.0}$, ≤ 2000 °C) [648, 660], 245 kJ mol^{-1} (hot-pressed $NbC_{0.99}$, 1600–2000 °C) [649], 275 kJ mol^{-1} (sintered $NbC_{0.80}$, 1600–2950 °C) [650]; (**c**) powder sintering densification – 265–320 kJ mol^{-1} ($NbC_{0.72}$, specific surface area 0.55–4.4 $m^2\ g^{-1}$, 1330–2150 °C) [659, 671], 260–310 kJ mol^{-1} ($NbC_{0.92}$, specific surface area 0.65–4.8 $m^2\ g^{-1}$, 1330–2150 °C) [659, 671], 290–310 kJ mol^{-1} ($NbC_{\sim 1.0}$ with carbon excess, 1300–1400 °C) [727, 731], 500–560 kJ mol^{-1} ($NbC_{\sim 1.0}$ with carbon excess, 2400–2700 °C) [727, 731]; (**d**) powder hot-pressing densification – 175 kJ mol^{-1} ($Nb_2C_{\sim 1.0}$, mean particle size 10–25 μm, 2300 °C) [833], 255 kJ mol^{-1}

(continued)

Table 4.22 (continued)

($NbC_{0.72}$, mean particle size 10–25 μm, 2300 °C) [833], 295 kJ mol^{-1} ($NbC_{0.82}$, mean particle size 10–25 μm, 2300 °C) [833], 325 kJ mol^{-1} ($NbC_{0.92}$, mean particle size 10–25 μm, 2300 °C) [833], 360 kJ mol^{-1} ($NbC_{0.99}$, mean particle size 10–25 μm, 2300 °C) [833]; (**e**) internal friction peaks characteristics (Marx-Wert formula) – 380 kJ mol^{-1} (sintered $NbC_{0.91}$, 1700 °C) [705]; data on creep – *see* Sect. 4.4 (Table 4.13), *see also* Sect. 4.5 (Table 4.16)

[b]On the near-surface effect in ^{14}C self-diffusion in NbC_{1-x} *see* [632]

[c]There are several evidences else that the layer growth rates of mono- and semicarbide phases in the deposited niobium as well as some other metallic thin films (coatings) are far greater (lower activation energy values) than those in ordinary bulk metals; more likely it occurs due to the greater contribution of grain-boundary diffusion as fine grain sizes, microporosity and higher dislocation density are inherent to the deposited films (coatings) [673, 674]

[d]The growth rate constants increase significantly in the presence of pulsed DC current with approximate density ~0.7 kA cm^{-2}, while the activation energies of formation of the NbC_{1-x} and β-$Nb_{2+x}C$ layers are independent on the presence of the current [676]

[e]The parameters probably contain significant contribution of surface and grain-boundary diffusion due to the rather high porosity of the materials

[f]At heating rates >10,000 K s^{-1} the sharp increase of carbidization rate occurs; it is caused by the formation of non-equilibrium liquid phase at temperatures noticeably lower than the melting point of the lowest eutectic in Nb-C system [677]

[g]Temperature dependences of the diffusion coefficients are the same for all the values of x in the indicated interval

Table 4.23 The interaction of near-stoichiometric niobium monocarbide NbC_{1-x} materials with some common chemical reagents in aqueous solutions [1, 62–64, 67, 82, 217, 514, 533, 684–687, 696]

Reagent, formula (density or concentration of aqueous solution)[a]	Treatment conditions		Character of interaction[b]
	Temperature, °C	Exposure time, h	
HCl (1:1)[c]	20	24	No decomposition
	110–112	2	Decomposes up to ~1%
HCl (d = 1.19)	20	24	No decomposition
	120	2	Decomposes up to ~4%
0.05M H_2SO_4	20	1–600	Corrosion rate (coatings on steel) – 2.1×10^{-5} mg cm^{-2} s
H_2SO_4 (1:4)	20	24	No decomposition
	112–115	2	Decomposes up to ~2%
H_2SO_4 (d = 1.84)	20	24	No decomposition
	260	2	Decomposes up to ~93% (36.3% amorphous C) with the formation of stable oxysulfates; composition of released gases: CO_2 79.0, CO 19.7 and H_2 1.3 mol.% (CO_2 and CO are products of the oxidation of amorphous C)[d]
	265	1.5	Decomposes up to ~81% (13.0% amorphous C) with the formation of stable oxysulfates; composition of released gases: CO_2 92.3, CO 5.4 and H_2 2.3 mol.% (CO_2 and CO are products of the oxidation of amorphous C)[d]
	280	1.5	Decomposes up to ~92% (47.0% amorphous C) with the formation of stable oxysulfates; composition of released gases: CO_2 73.8, CO 24.3 and H_2 1.9 mol.% (CO_2 and CO are products of the oxidation of amorphous C)[d]
	290–310	1	Decomposes completely with the formation of stable oxysulfates; composition of released gases: CO_2 82.9, CO 13.7 and H_2 3.4 mol.% (CO_2 and CO are products of the oxidation of amorphous C, solution is transparent finally)[d]
HNO_3 (1:1)	20	24	Partial dissolution (deposition of salts, partial hydrolysis)
	105	2	Partial dissolution (deposition of salts, partial hydrolysis)
HNO_3 (d = 1.43)	20	24	No decomposition
	110–120	2	Partial dissolution (deposition of salts, partial hydrolysis)

(continued)

Table 4.23 (continued)

Reagent, formula (density or concentration of aqueous solution)[a]	Treatment conditions		Character of interaction[b]
	Temperature, °C	Exposure time, h	
HF (d = 1.15)	20	24	Decomposes
H_3PO_4 (1:3)	20	24	Decomposes up to ~10%
	≥100	2	Decomposes up to ~1%
H_3PO_4 (d = 1.70)	20	24	No decomposition
	120–300	2	Decomposes very slightly
$H_2C_2O_4^e$ (saturated solution)	20	24	Decomposes up to ~1%
	100–104	2	Decomposes up to ~1%
NaOH (10%)	20	24	Decomposes up to ~1%
NaOH (20%)	20	24	Decomposes up to ~1%
	110	2	No decomposition
NaCl (3%)	20	1–600	Corrosion rate (coatings on steel) – 2.1×10^{-6} mg cm^{-2} s^{-1}
Na_2CO_3 (10%)	20	1–600	Corrosion rate (coatings on steel) – 2.7×10^{-7} mg cm^{-2} s^{-1}
KOH (30%)	20	1–600	Corrosion rate (coatings on steel) – 4.0×10^{-7} mg cm^{-2} s^{-1}
H_2O (technical)	20	1–600	Corrosion rate (coatings on steel) – 1.0×10^{-6} mg cm^{-2} s^{-1}
H_2O_2 (30%)[f]	110	1.0–1.5	Decomposes completely
NH_4F (5%)	–	–	Decomposes up to ~8%
$(NH_4)_2S_2O_8$ (20–25%)	110	1	Decomposes up to ~10–15%
HCl (1:1) + H_2SO_4 (1:4)	110	1	No decomposition
3HCl (1:1) + HNO_3 (1:2)	110	1	No decomposition
3HCl (d = 1.19) + HNO_3 (d = 1.43)	20	24	Decomposes up to ~8%
	105	2	Complete dissolution with hydrolysis
HCl (d = 1.19) + H_2O_2 (30%)	110	1	Decomposes up to ~12–15%
HCl (d = 1.19) + $(NH_4)_2S_2O_8$ (25%)	110	1	Decomposes up to ~3–5%
HCl (d = 1.19) + Br_2 (HBrO, HBr) (saturated solution)	–	–	Decomposes up to ~12%
H_2SO_4 (d = 1.84) + HNO_3 (d = 1.43)	20	24	No decomposition
	140	2	Decomposes up to ~78%
H_2SO_4 (1:4) + H_3PO_4 (1:3)	≥150	2	Decomposes up to ~1%
H_2SO_4 (d = 1.84) + H_3PO_4 (d = 1.70)	20	24	Decomposes up to ~9%
	240	2	Decomposes completely
H_2SO_4 (1:4) + $H_2C_2O_4^e$ (saturated solution)	20	24	No decomposition
	180	2	Decomposes up to ~5%
H_2SO_4 (d = 1.84) + $H_2C_2O_4^e$ (saturated solution)	20	24	No decomposition
	≥100	2	Decomposes up to ~5%

(continued)

Table 4.23 (continued)

Reagent, formula (density or concentration of aqueous solution)[a]	Treatment conditions		Character of interaction[b]
	Temperature, °C	Exposure time, h	
H_2SO_4 (1:4) + H_2O_2 (30%)	110	1	Decomposes up to ~80–87%
H_2SO_4 (1:4) + $(NH_4)_2S_2O_8$ (25%)	110	1	Decomposes up to ~35–55%
H_2SO_4 (1:1) + $(NH_4)_2S_2O_8$ (25%)	110	1	Decomposes up to ~78%
(1–4)HNO_3 (d = 1.43) + HF (d = 1.15)[g]	20	24	Decomposes completely
	110	0.5–2.0	Decomposes completely
HNO_3 (d = 1.43) + NH_4F (5%)	110–120	0.5	Decomposes completely; composition of released gases: CO_2 95–97, CO (traces) and H_2 2–3 mol.% (no precipitated C, solution is transparent)
H_2O_2 (30%) + $H_2C_2O_4^e$ (50%)	110	1	Decomposes completely
H_2O_2 (30%) + $H_2C_4H_4O_6^h$ (50%)	110	1	Decomposes completely
H_2O_2 (30%) + $H_3C_6H_5O_7^i$ (50%)	110	1	Decomposes completely
H_2O_2 (30%) + $H_4C_{10}H_{12}N_2O_8^j$ (saturated solution)	110	1	Decomposes completely
$(NH_4)_2S_2O_8$ + $H_4C_{10}H_{12}N_2O_8^j$	–	–	Decomposes up to ~2%
$(NH_4)_2S_2O_8$ + NH_4F	–	–	Decomposes up to ~20%
4NaOH (20%) + Br_2 (HBrO, HBr)	20	24	No decomposition
	105	2	Decomposes up to ~16% (partial dissolution with hydrolysis)
4NaOH (20%) + H_2O_2 (30%)	20	24	Decomposes up to ~29%
	112	2	Decomposes up to ~12%
4NaOH (20%) + $K_3[Fe(CN)_6]$ (10%)	20	24	Partial dissolution (hydrolysis)
	110	2	Partial dissolution (hydrolysis)

[a]All the ratios are given in volume fractions

[b]When it is not indicated specially, the character reported is related to the powders with mean particle size of 40–50 μm

[c]Anodic oxidation in hydrochloric acid HCl results in the partial oxidation of carbon in the carbide with the formation of gaseous products CO and CO_2 and simultaneous release of hydrogen peroxide H_2O_2; the latter one participates in the formation of water-soluble complexes of Nb, which destabilize the protective oxide (with residual carbon) film on the carbide anode [688]

[d]Represented by the reaction equation: $2NbC_{1-x} + 4H_2SO_4 + 5SO_3 = Nb_2O(SO_4)_4 + 2(1 - x)C + 5SO_2 + 4H_2O$

[e]Oxalic acid

[f]In the mixtures of H_2O_2 (30%) with citric acid $H_3C_6H_5O_7$, oxalic acid $H_2C_2O_4$, tartaric acid $H_2C_4H_4O_6$, ethylenediaminetetraacetic acid $C_{10}H_{16}N_2O_8$ (EDTA) and ammonium fluoride NH_4F decomposes completely in 20 min [533]

[g]Recommended chemical etching agents for NbC_{1-x}: (**a**) 10 ml HNO_3 (d = 1.43) + 3 ml HF (d = 1.15) with 0.5–1.0 min exposure at room temperature (for cleavage planes and rough surfaces) [679]; (**b**) 50 ml H_2SO_4 (d = 1.84) + 20 ml HNO_3 (d = 1.43) + 10 ml HF (d = 1.15) with 5 min exposure at 0 °C (for the electrochemical treatment (anodization) of polished sections, cathode – graphite or Pt, voltage – 30 V, current density – 0.8–1.0 A mm^{-2}, continuing addition of HF solution, followed by chemical etching in accordance to *a*) [679]; (**c**) 10 g $K_3[Fe(CN)_6]$ + 2–10 g KOH/NaOH + 100 ml H_2O (Murakami's reagent for metallographic analysis) [683, 692]; (**d**) boiling mixture of H_2SO_4 (d = 1.84) + $(NH_4)_2S_2O_8$ [680]; (**e**) H_2NSO_3H (20% sulfamic acid) with 1 min exposure (for thin films on different substrates) [681, 682]; (**f**) HF (d = 1.15) + 40% NH_4F (BHF solution) with 4 s exposure (for thin films on different substrates) [681, 682]; (**g**) 50 ml HCl (d = 1.19) + 5 ml HNO_3

(continued)

Table 4.23 (continued)

(d = 1.43) + 50 ml H_2O (V2A etchant) with 1 min exposure at 40 °C [683]; (**h**) 40 ml HCl (d = 1.19) + 10 ml HNO_3 (d = 1.43) + 1 ml HF (d = 1.15) with 4 min exposure at 70 °C (for dislocation etching) [297]; (**i**) HNO_3 (d = 1.43) + 2–3% Na_2SiF_6 with 4 min exposure at room or cooler temperatures [692]; (**j**) 1 ml HNO_3 (d = 1.43) + 1 ml HF (d = 1.15) + 6 ml H_2O with 4 s exposure (for dislocation electroetching) [297]

[h]Tartaric acid

[i]Citric acid

[j]Ethylenediaminetetraacetic acid (EDTA)

In comparison with other ultra-high temperature materials the summarized data on the chemical behaviour of niobium carbides are given in Addendum.

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Chapter 5
Zirconium Monocarbide

5.1 Structures

Practically zirconium forms with carbon the only one chemical compound (*see* also section C–Zr in Table I-2.13) – zirconium monocarbide ZrC_{1-x} with extremely broad homogeneity range, apart from low-temperature (<860–1080 °C) ordered structures of $Zr_{2\pm x}C$ (*Fd(–3)m*, *R(–3)m*, *Pnma*, *P3m1*, *I4$_1$/amd*, *Pbcn*, *P4/mmm*), $Zr_3C_{2\pm x}$ (*C2/m*, *P(–3)m1*, *I4/mmm*, *Immm*), $Zr_4C_{3\pm x}$ (*C2/c*, *P(–3)m1*, *Pm(–3)m*), $Zr_5C_{4\pm x}$ (*P(–1)*, *C2/m*, *I4/m*), $Zr_6C_{5\pm x}$ (*C2/m*, *C2/c*, *P3$_1$12*), $Zr_7C_{6\pm x}$ (*R(–3)*) and $Zr_8C_{7\pm x}$ (*P4$_3$32*, *Fm(–3)m*) as well as molecular clusters Zr_8C_{12}, $Zr_{13}C_{22}$, $Zr_{14}C_{23}$, $Zr_{18}C_{29}$, $Zr_{22}C_{35}$, including endohedral $C@Zr_8C_{12}$, which are not confirmed sufficiently in literature [1–14, 696, 702, 1029, 1080, 1143, 1376, 1391, 1408, 1409, 1420]. The ordered structures should be in thermodynamic equilibrium with disordered monocarbide phase; however, the latter one can exist as a metastable phase for an infinitely long time because the diffusion rate at lower temperatures is extremely small [2, 8]. A high-temperature partial variant of zirconium–carbon phase diagram is given in Fig. 5.1, and the structural features of zirconium monocarbide ZrC_{1-x} are presented in Table 5.1. The C/Zr radii ratio, calculated on the basis of Pauling's atomic size of Zr (0.1597 nm, CN = 12) is 0.483 [11, 74]; the ratio of Zr radii (in nm) in Me/MeC is 0.161/0.166 (2.8% expansion of Zr atoms in carbide) [17]. The phase transformation of ZrC_{1-x} from NaCl type (*Fm(–3)m*) to CsCl type (*Pm(–3)m*) structure under very high pressures (~100–500 GPa) was predicted theoretically [18–25, 466, 716, 1372].

The variation of the lattice parameter of zirconium monocarbide ZrC_{1-x} phases with carbon content (or deviation from the stoichiometry) within the homogeneity range is shown in Fig. 5.2; in a modified form the following equation described this relationship was proposed by Senczyk (for $0 \leq x \leq 0.5$) [1476]:

$$a = 0.4701 - 0.00665x - 0.00185x^2, \tag{5.1}$$

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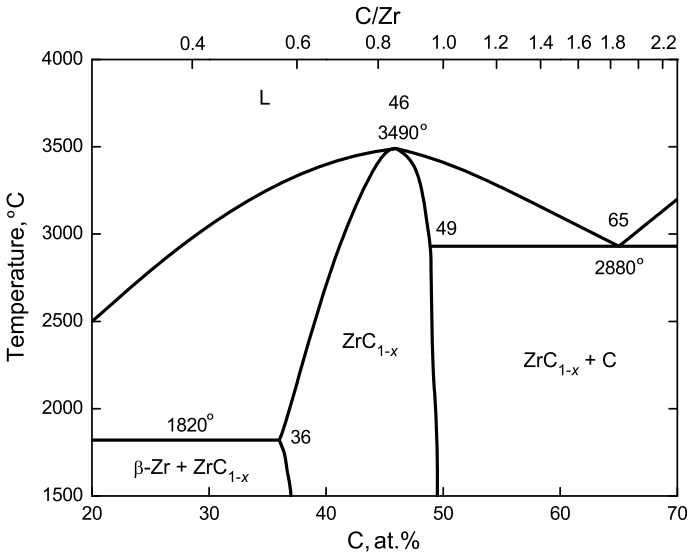

Fig. 5.1 High-temperature partial variant of zirconium – carbon equilibrium phase diagram [2, 4, 8, 10, 11, 15, 16, 26, 27, 43, 72–76, 80, 96–99, 105, 146, 201, 267–269, 600, 1388, 1389, 1450, 1539]

and Gusev (for $x \leq 0.355$) [160]:

$$a = 0.46965 + 0.00697x - 0.02018x^2, \tag{5.2}$$

where a is lattice parameter, nm and x is the value of index in ZrC_{1-x} formula. The wide spread of data available in literature for the relationships between the lattice parameter and composition of ZrC_{1-x} mainly could be explained by some difficulties in the contamination control of phase composition; the considerable solubilities of oxygen and nitrogen, which are always present in ZrC_{1-x}, result in the formation of oxycarbide $ZrC_{1-x}O_y$, carbonitride $ZrC_{1-x}N_z$ and more commonly oxycarbonitride $ZrC_{1-x}N_zO_y$ phases (*see also* sections C–N–Zr, C–N–O–Zr and C–O–Zr in Table I-2.14) with various deviations from the stoichiometry and noticeable decrease in the magnitude of lattice parameter observed due to the direct substitution of carbon for oxygen and/or nitrogen [26, 27, 44, 151, 161–166, 598, 1077]. On the basis of various literature values for a range of oxygen and/or nitrogen impurity contents Mitrokhin et al. [71] established the quantitative relationship between lattice parameter and composition for zirconium monocarbide (oxycarbonitride in practice) $ZrC_{1-x}(N,O)_y$ phases, which looks in a modified form as

$$a = 0.4696 + 0.0074x - 0.0208x^2 - 0.080xy, \tag{5.3}$$

Table 5.1 Structural properties (crystal structure, density) of zirconium monocarbide phase

Formula	Crystal structure					Density[c], g cm^{-3}	Reference
	System	Type	Space group	Lattice parameter[a] a, nm	Z[b]		
ZrC_{1-x}	Cubic	NaCl	*Fm(–3)m*	0.4652[d]	4	–	[79]
				0.4656[e]	4	–	[78]
				0.4665[f]	4	–	[44, 162]
				0.4669	4	6.74	[150]
				0.4670	4	6.73	[148]
				0.4670	4	6.47	[36, 139]
				0.4673	4	–	[37]
				0.4675[g]	4	–	[146]
				0.4676[h]	4	–	[44, 162]
				0.4677[i]	4	–	[61]
				0.4680[j,k]	4	–	[57, 153]
				0.4681[l]	4	–	[44, 161]
				0.4682[m,n]	4	–	[53, 72, 155]
				0.4683[o]	4	6.66	[32, 47, 100]
				0.4683	4	6.65	[130]
				0.4683	4	6.64	[127, 128]
				0.4683[p]	4	6.31	[154]
				0.4683[q]	4	–	[78, 237]
				0.4686	4	6.66	[140]
				0.4687	4	6.66	[41, 143, 236]
				0.4688	4	–	[30, 31, 42, 45]
				0.4688[r,s,t]	4	–	[44, 62, 81, 163]
				0.4689[u,v]	4	–	[38, 39, 62, 68]
				0.4690[q,w]	4	–	[40, 78, 844]
				0.4691	4	6.64	[141]
				0.4691[x]	4	6.33	[10]
				0.4691[x]	4	6.29	[43]
				0.4691[q,y]	4	–	[66, 242, 243]
				0.4692[q,z]	4	–	[77, 88]
				0.4692[q]	4	6.64	[117]
				0.4692	4	6.52	[131]
				0.4693[a1,a2,a3]	4	–	[54, 65, 83, 89]
				0.4693[o,a4]	4	–	[91, 132, 153]
				0.4693	4	6.63	[124, 133, 147]
				0.4693	4	6.60	[136]
				0.4693	4	6.56	[135]
				0.4693[a5]	4	6.46	[844]

(continued)

Table 5.1 (continued)

Formula	Crystal structure					Density[c], g cm^{-3}	Reference
	System	Type	Space group	Lattice parameter[a] a, nm	Z[b]		
				0.4694[a3,a6,a7]	4	–	[57, 72, 101, 102]
				0.4694[a8]	4	6.63	[111, 122, 129]
				0.4694[a9]	4	6.62	[844]
				0.4694[b1]	4	6.61	[109]
				0.4694[b2]	4	6.35	[85]
				0.4694[q]	4	–	[146, 1482]
				0.4695[a6,b3,b4]	4	–	[26, 27, 69, 94]
				0.4695[b5]	4	6.62	[844]
				0.4695	4	6.60	[106–108]
				0.4695[b6]	4	6.50	[844]
				0.4695[b7]	4	–	[156]
				0.4696[b8,b9]	4	–	[44, 60, 71, 95]
				0.4696	4	6.62	[145, 149]
				0.4696[a6]	4	6.60	[120, 121]
				0.4697[c1]	4	6.73	[1, 3, 4, 26–29]
				0.4697[c2,c3,c4]	4	–	[59, 64, 84]
				0.4697[o,c5,c6]	4	–	[91, 97–99, 200]
				0.4697[c7]	4	6.62	[110, 116, 126]
				0.4697	4	6.59	[91, 118, 132]
				0.4697	4	6.58	[123]
				0.4698	4	6.73	[63, 67]
				0.4698[c8]	4	6.66	[2, 10, 33–35, 52]
				0.4698[c9]	4	6.62	[58]
				0.4698[b3]	4	6.59	[11, 48, 55, 112]
				0.4698[d1]	4	6.57	[43, 46, 53, 72]
				0.4698[d2]	4	6.56	[56]
				0.4698[d3]	4	6.55	[87]
				0.4698[d4]	4	6.51	[58]
				0.4698	4	6.50	[119]
				0.4698[q,d5,d6]	4	–	[89, 152, 156]
				0.4698[d5,d7]	4	–	[157, 186]
				0.4699[d8,d9]	4	–	[49, 50, 61, 62]
				0.4699[q]	4	6.61	[77, 114, 115]
				0.4699[e1,e2,e3]	4	–	[83, 84, 88]
				0.4699[e4]	4	6.60	[85]
				0.4699	4	6.59	[142]
				0.4699[p,e5,e6]	4	–	[90, 92, 200]
				0.4699	4	6.54	[125]
				0.4700	4	6.60	[113, 144]
				0.4700[e7]	4	6.58	[82]
				0.4700[e8,e9]	4	–	[51, 81, 86]

(continued)

Table 5.1 (continued)

Formula	Crystal structure					Density[c], g cm^{-3}	Reference
	System	Type	Space group	Lattice parameter[a] a, nm	Z[b]		
				0.4700[f1]	4	–	[156]
				0.4701[c7,f2,f3]	4	–	[88, 89, 152]
				0.4701[f4]	4	–	[415]
				0.4701	4	6.58	[134]
				0.4701[a4]	4	6.56	[154]
				0.4702[f5,f6,f7]	4	–	[10, 70, 81, 84]
				0.4702[f8]	4	6.50	[85]
				0.4702[f9,g1]	4	–	[93, 200]
				0.4703	4	6.52	[137, 138]
				0.4708[g2]	4	–	[175]
				0.475–0.478[g3]	4	–	[103, 104, 191]

[a]When it is not indicated specially, value reported is for near-stoichiometric composition
[b]Number of formula units per lattice cell
[c]Calculated from XRD or neutron diffraction patterns
[d]Carbon content – 36.6 at.%
[e]Carbon content – 27.6 at.% (?)
[f]Taking into account O contamination, it is corresponding to $ZrC_{0.65}O_{0.03}$
[g]Carbon content – 35.0 at.%
[h]Taking into account O contamination, it is corresponding to $ZrC_{0.90}O_{0.04}$
[i]High purity materials with carbon content – 42.5 at.%
[j]Carbon content – 39.4 at.%
[k]Carbon content – 37.5 at.%
[l]Taking into account O contamination, it is corresponding to $ZrC_{0.64}O_{0.05}$
[m]Carbon content – 35.9 at.% (in equilibrium with α-Zr)
[n]Carbon content – 36.5 at.%
[o]Carbon content – 49.0 at.%
[p]Carbon content – 36.7 at.%
[q]Carbon content – ~50 at.%
[r]Carbon content – 49.7 at.% (combined), non-combined C – 0.2%, O – 0.1%, N – 0.1%
[s]Carbon content – 41.9 at.%, O – 0.77%
[t]Taking into account O contamination, it is corresponding to $ZrC_{0.90}O_{0.05}$
[u]Carbon content – 40.5 at.% (combined), non-combined C – absent, O – 0.3%, N – 0.1%
[v]Calculated by all-electron full-potential linearized augmented plane-wave (FLAPW) method
[w]Prepared by self-propagating high-temperature synthesis (SHS) with charged C/Zr ratio – 0.7, carbon content – 37.5 at.%
[x]Carbon content – 35.5–36.7 at.% (in equilibrium with α-Zr)
[y]Interatomic distances: Zr–Zr and C–C – 0.33171 nm, Zr–C – 0.23455 nm
[z]Carbon content – 35.5 at.%
[a1]Reported thermal increment is 6.8×10^{-6} K^{-1}
[a2]Carbon content – 38.7 at.%, O – 0.32%, N – 0.014%
[a3]Carbon content – 38.5 at.%
[a4]Carbon content – 39.0 at.%
[a5]Prepared by SHS with charged C/Zr ratio – 0.8, carbon content – 41.5 at.%
[a6]Carbon content – 49.2 at.%
[a7]Related to the carbide phase (contents O – 0.6–1.0%, N – 0.3–0.5%) in equilibrium with graphite in two phase carbide-carbon compositions
[a8]99.9% purity materials, carbon content 49.8 at.%

(continued)

Table 5.1 (continued)

[a9]Prepared by SHS with charged C/Zr ratio – 2.0, carbon content – 49.7 at.%
[b1]Carbon content – 49.5 at.%, O – 0.15%, Hf – 2.5%
[b2]Carbon content – 39.0 at.%, O – 0.13%
[b3]Carbon content – 38.3 at.%
[b4]Calculated on the basis of density-functional theory (DFT) with generalized gradient approximation (GGA)
[b5]Prepared by SHS with charged C/Zr ratio – 1.5, carbon content – 49.7 at.%
[b6]Prepared by SHS with charged C/Zr ratio – 1.0, carbon content – 45.7 at.%
[b7]Experimental data extrapolated from $ZrC_{0.79}O_{0.19}$ measurements to an oxygen-free phase
[b8]Carbon content – 48.7 at.%
[b9]Experimental data extrapolated to $x = y = 0$ for $ZrC_{1-x}(O,N)_y$ compounds
[c1]Carbon content – 49.2–49.7 at.%; minimal interatomic distances: Zr–Zr and C–C – 0.332 nm, Zr–C – 0.235 nm
[c2]Experimental data extrapolated to $y = 1$ for ZrC_yN_{1-y} compounds
[c3]For nanoparticles; taking into account O contamination, it is corresponding to $ZrC_{0.93}O_{0.02}$
[c4]Carbon content – 38.3 at.%, O – 0.005%
[c5]Carbon content – 45.1–46.8 at.%
[c6]Related to the carbide phase (O – 0.5%, N < 0.001%) in equilibrium with graphite (lattice parameter increases after the melting procedure of two phase carbide-carbon composition by 0.0002–0.0013 nm)
[c7]Carbon content – 46.2 at.%
[c8]Carbon content – 49.2 at.% (in equilibrium with graphite)
[c9]Taking into account N and O contaminations, it is corresponding to $ZrC_{0.88}N_{0.04}O_{0.02}$
[d1]Carbon content – 48.7–49.5 at.% (in equilibrium with graphite)
[d2]Experimental data extrapolated from $(Zr_{0.992}Hf_{0.008})C_{0.937}$ (contents: non-combined C – 0.26%, O < 0.15%, N < 0.01%) measurements to a Hf-free stoichiometric phase; minimal interatomic distance Zr–C – 0.2349 nm
[d3]Single crystal materials, carbon content – 48.7 at.%
[d4]Taking into account N and O contaminations, it is corresponding to $ZrC_{0.73}N_{0.03}O_{0.02}$
[d5]Chemical vapour deposited whiskers, carbon content – 40.3 at.%, O – 0.05%
[d6]Experimental data extrapolated from $ZrC_{0.83}O_{0.15}$ measurements to an oxygen-free phase
[d7]Nanoparticles with average size ~7 nm, carbon content – 49.9 at.%
[d8]High purity materials with carbon content – 49.5 at.%
[d9]Carbon content – 47.6 at.% (combined), non-combined C – absent, O – 0.1%, N – 0.1%
[e1]Carbon content – 49.5 at.% (combined), non-combined C – 0.5%, O – 0.04%, N – 0.003%
[e2]Carbon content – 49.7 at.%, O – 0.02%
[e3]Carbon content – 47.9 at.%
[e4]Carbon content – ~50 at.%, O – 0.05%
[e5]Carbon content – 48.2 at.%
[e6]Carbon content – 40.0 at.%
[e7]Carbon content – 49.1 at.% (combined), non-combined C – 0.4%, O – 0.07–0.09%, N – 0.1%
[e8]Carbide phase in equilibrium with graphite (content O – 0.36%)
[e9]Carbon content – 49.3 at.% (combined), non-combined C – 0.15%, O – 0.62%, N – 0.018%
[f1]Experimental data extrapolated from $ZrC_{0.87-0.89}O_{0.007-0.070}$ measurements to an oxygen-free phase
[f2]Carbon content – 43.5 at.%
[f3]Chemical vapour deposited whiskers, carbon content – 43.8–45.3 at.%, O – 0.05%
[f4]Thin films (2 μm thickness) prepared by activated reactive evaporation
[f5]Carbon content – 47.6 at.%, O – 0.04%
[f6]Carbon content – 45.1–46.6 at.%, O – 0.02–0.07%, N – 0.004–0.005%
[f7]Carbon content – 45.4 at.%; the maximal value measured for the phase lattice parameters in homogeneity region
[f8]Carbon content – 46.5 at.%, O – 0.06%
[f9]Carbon content – 47.4 at.%
[g1]Carbon content – 45.0 at.%
[g2]Powders with nanoscale dimensions ~5 nm
[g3]Related to nanocrystalline phase (~2–12 nm grains) in contact with amorphous carbon

where a is lattice parameter, nm and x is the value of index in ZrC_{1-x} (or oxycarbonitride $ZrC_{1-x}(N,O)_y$) formula ($0 < x \leq 0.38$) and y is the (O + N)/Zr atomic ratio ($y < 0.3$). Some examples of the effects of temperatures and oxygen impurity atoms on the lattice parameters of ZrC_{1-x} monocarbide phases are demonstrated in Figs. 5.3 and 5.4; the examples of impurities effect on the completeness of metal and non-metal sublattices in ZrC_{1-x} ($ZrC_{1-x}N_zO_y$) are given in Table 5.2.

Like most other transition metal carbides, metal and carbon atoms in the zirconium monocarbide ZrC_{1-x} lattice are bound by ionic and covalent (C–Zr) and metallic (Zr–Zr) bonds; the latter bonds are known to be significantly weaker than metal-to-carbon bonds. Although ZrC_{1-x} is a polar covalent compound with considerable charge transfer from metal to carbon atoms [167, 168], namely the strong covalent character of chemical bonding in ZrC_{1-x} determines the high value of C–Zr bond energy (enthalpy), which is about 5.81 ± 0.3 eV [96, 1384], or 5.14 ± 0.4 eV [746, 1385, 1386].

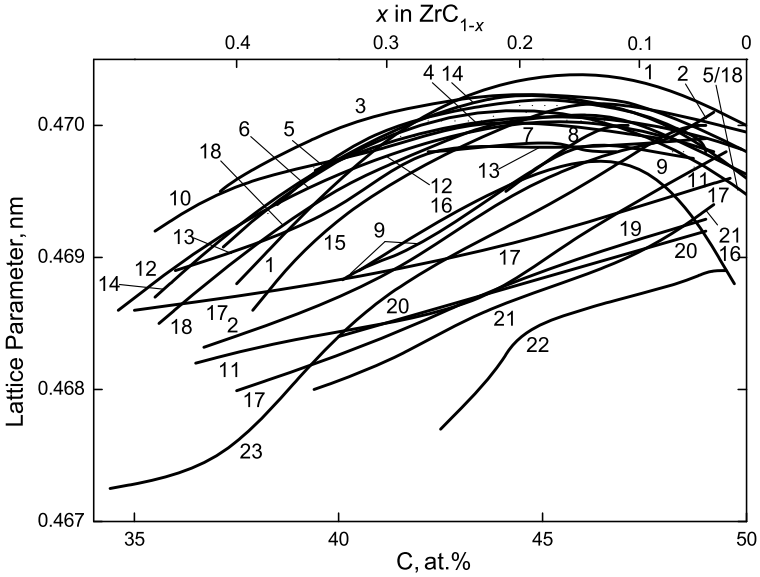

Fig. 5.2 Lattice parameter of ZrC_{1-x} as a function of phase composition: 1 – [160], 2 – synthesized from elements [154], 3 – arc-melted and annealed in vacuum [84], 4 – chemical vapour deposited whiskers [152], 5 – polynomial approximation [26, 27], 6 – synthesized by reacting Zr hydride with carbon and annealed [80], 7 – $ZrC_{1-x}N_{0.02-0.04}O_{0.02-0.03}$ synthesized from elements in H_2 [58], 8 – measurement extrapolation to oxygen free phases [156], 9 – [53], 10 – [88], 11 – [155], 12 – [10], 13 – arc-melted and annealed [158], 14 – polynomial approximation extrapolated to oxygen free phases [71, 74], 15 – [75], 16 – contents O – 0.1–0.3%, N – 0.1% [62, 1311], 17 – prepared by the carbothermal reduction of Zr oxide [98], 18 – synthesized by different methods [546, 1449], 19 – [153], 20 – [159], 21 – synthesized from elements in vacuum [57, 513, 877, 1308], 22 – contents O, N – 0.05–0.15% [61], 23 – [1477, 1478] (1, 5, 11, 13–14, 18 – on the basis of several sources)

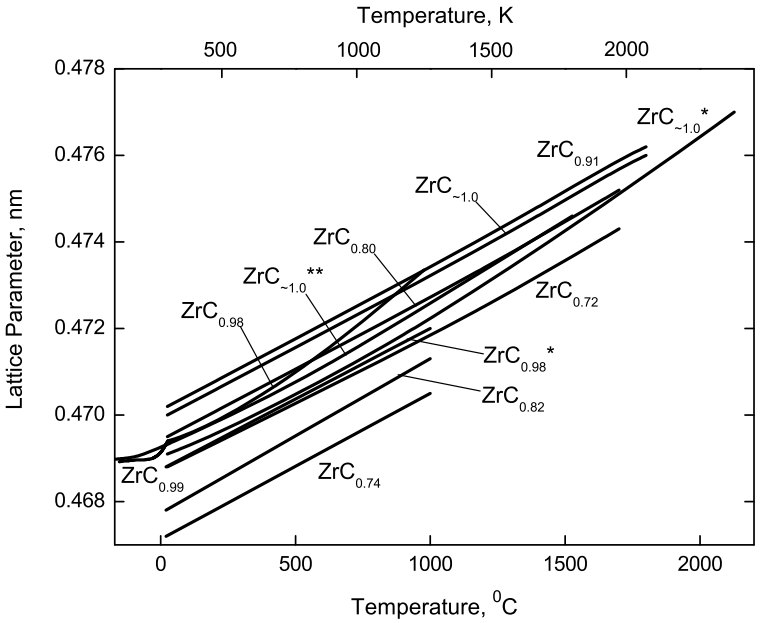

Fig. 5.3 Lattice parameter of ZrC_{1-x} as a function of temperature for the phases having different deviations from the stoichiometry ($x \approx 0$ (in equilibrium with carbon), $x = 0.09$, $x = 0.20$ and $x = 0.28$, synthesized by the reaction of Zr hydride with carbon in vacuum, content O – 0.04–0.77% [81]; $x \approx 0$*, on the basis of several sources [242]; $x \approx 0$**, prepared by the carbothermic reduction of Zr oxide, content O < 0.01%, measured by neutron diffraction [362]; $x = 0.01$, 99.9% purity [111]; $x = 0.02$*, contents: O – 0.15%, Hf – 2.5% [109]; $x = 0.02$*, $x = 0.18$ and $x = 0.26$ [1350])

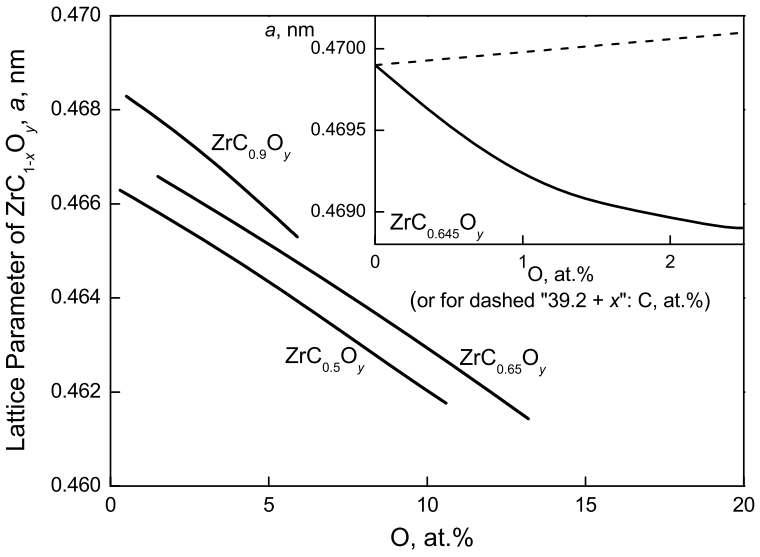

Fig. 5.4 Effect of oxygen atoms on the lattice parameter of $ZrC_{1-x}O_y$ oxycarbide phases formed: $ZrC_{0.9}O_y$, $ZrC_{0.65}O_y$ and $ZrC_{0.5}O_y$ [26, 27] (*Inset* – for $ZrC_{0.645}O_y$ phase, the dashed line shows the change of parameter if carbon atoms were added to the vacancies instead of oxygen [83, 84])

Table 5.2 The completeness of zirconium monocarbide (oxycarbonitride) ZrC_{1-x} ($ZrC_{1-x}N_zO_y$) sublattices (with various impurity contents) [3, 44, 58, 151, 1445]

Phase composition	Fraction of structural defects in sublattices	
	metal	non-metal
$ZrC_{0.88}N_{0.04}O_{0.02}$	–	0.06
$ZrC_{0.85}O_{0.15}$	0.015	0.015
$ZrC_{0.79}O_{0.21}$	0.0225	0.025
$ZrC_{0.74}N_{0.04}O_{0.02}$	–	0.20
$ZrC_{0.74}O_{0.26}$	0.035	0.035
$ZrC_{0.73}O_{0.19}$	0.025	0.1025
$ZrC_{0.70}O_{0.24}$	0.0325	0.09
$ZrC_{0.63}O_{0.33}$	0.055	0.0925
$ZrC_{0.54}O_{0.19}$	0.02	0.2825
$ZrC_{0.50}O_{0.09}$	0	0.41
$ZrC_{0.50}O_{0.21}$	0.075	0.3425

Table 5.3 The DFT-calculated values of unstable stacking fault energies and total energies required to shear for the crystals of near-stoichiometric zirconium monocarbide ZrC_{1-x} along different slip systems [174]

Slip system	Unstable stacking fault energy[a], eV nm^{-2}	Shear total energy[b], eV nm^{-3}
(0 0 1) <110>	16.4	77
(1 $\bar{1}$ 0) <110>	14.1	331
(1 $\bar{1}$ 1) <110>	15.6	305

[a]Resistance to dislocation nucleation processes [182]
[b]Measure of ideal shear stress

It was determined mainly by the hardness measurements on single crystal materials that the preferred slip systems for near-stoichiometric ZrC_{1-x} phases are (110)<1$\bar{1}$0> – at moderate temperatures, (100)<1$\bar{1}$0>, (110)<1$\bar{1}$0> and (111)<1$\bar{1}$0> – at elevated and higher temperatures, and the combination of (110)<1$\bar{1}$0> and (111)<1$\bar{1}$0> systems – for the intermediate regions [11, 26, 27, 169–173, 572, 658, 671, 672, 1470, 1503, 1504]. However, Kiani et al. [174, 1407], using *in situ* electron microscopy based nanomechanical testing under compression jointly with density functional theory calculations, identified the (001) <110> slip system at quasi-room temperature as an active system operating in ZrC (111) single-crystalline nanopillars; the energy values, DFT-calculated by them, which characterize several different possible slip systems in near-stoichiometric ZrC_{1-x}, are given in Table 5.3. The minimal Burgers vector of near-stoichiometric monocarbide ZrC_{1-x} (½ <110>) b = 0.332–0.334 nm [4, 52, 659]. For parameters of formation and migration of lattice point defects (vacancies and interstitial atoms) in ZrC_{1-x} *see* Sect. 5.5 (Table 5.19).

Recently, due to massive progress in nanotechnology, many nanostructures based on ZrC_{1-x}, including various (in types) nanopowders and nanoparticles (with

mean sizes from 5 to 500 nm) [64, 148, 175–181, 185–188, 375, 987, 1547], nanofibres (40–700 nm in diameter, few micrometers in length) [183, 184, 194], whiskers (1–20 μm diameter, 0.1–7 mm length) grown along <100> and <110> crystal directions [152, 157, 878, 1464, 1548], various nanocrystalline thin films/coatings (up to 0.6 nm in thickness) [103, 104, 144, 193, 198, 580, 581, 663, 729, 1060, 1491, 1493, 1495], have been synthesized; 2D-molecular $Zr_{n+1}C_n$ (or Zr_2CT_x, where T are functional groups OH, O, F) MXenes [189–192, 195–197, 1398, 1406, 1441, 1442] were synthesized and examined using first principles calculations too.

According to Storms [10], the X-ray density of zirconium monocarbide ZrC_{1-x} phase changes from 6.33 g cm^{-3} for $ZrC_{0.58}$ to 6.59 g cm^{-3} for $ZrC_{0.97}$. For the bulk density of poreless ZrC_{1-x} single-phase materials with various deviations from the stoichiometry Katoh et al. [96] proposed the following approximation function:

$$d = 6.61 - 0.88x + 0.32x^2, \tag{5.4}$$

where d is the bulk mass density, g cm^{-3} and x is the value of index in ZrC_{1-x} formula ($0 < x \leq 0.4$). The recommended value for the density of pure poreless near-stoichiometric zirconium monocarbide ZrC_{1-x} materials at room temperature is 6.60–6.65 g cm^{-3} [1, 4, 30, 32, 33, 37, 45, 199].

5.2 Thermal Properties

Zirconium monocarbide ZrC_{1-x} has one of the highest melting points of all the solid substances available. Within the homogeneity range of ZrC_{1-x} the melting point of the phase varies (*see* Fig. 5.1); the maximum temperature is pertaining to the nonstoichiometric composition $ZrC_{0.82\pm0.03}$ [2–4, 10, 11, 15, 16, 72–80, 98, 200, 201]. The general thermodynamic properties of near-stoichiometric zirconium monocarbide are summarized in Table 5.4. For the molar heat capacity of near-stoichiometric zirconium monocarbide ZrC_{1-x} $c_p = f(T, \text{K})$, J mol^{-1} K^{-1}, the following relationships were recommended in the literature:

in the range of temperatures from 40 to 1380 K [54]

$$c_p = 58.58 \exp(-132.1/T), \tag{5.5}$$

in the range of temperatures from 300 to 1200 K [250]

$$c_p = 47.70 + \left(5.941 \times 10^{-3}\right)T - \left(7.811 \times 10^{5}\right)T^{-2}, \tag{5.6}$$

in the range of temperatures from 300 to 3300 K [31, 42, 207]

$$c_p = 54.81 + \left(2.218 \times 10^{-3}\right)T - \left(1.105 \times 10^{7}\right)T^{-2}, \tag{5.7}$$

Table 5.4 General thermodynamic properties of near-stoichiometric zirconium monocarbide ZrC_{1-x}

Characteristics	Symbol	Unit	Value	Reference
Standard heat of formation (at 298.15 K)[a]	$-\Delta H^\circ_{298}$	kJ mol^{-1}	171.5	[230]
			182.9[b]	[376]
			184.5 ± 6.3	[31, 35, 46, 207, 217, 218, 223, 375, 635]
			190.0	[37]
			196.6 ± 13.0	[1, 208, 221, 224, 225, 241, 260, 272]
			198.4 ± 4.0	[264]
			198.7 ± 6.3	[231]
			199.4 ± 12.6	[32, 39, 42, 47, 48]
			200.0 ± 6.3	[219, 220, 251, 381]
			200.8 ± 20.9	[30, 33, 211, 229]
			201.1 ± 2.5	[261]
			202.0 ± 2.5[c]	[83, 262, 273]
			206.6 ± 10.0[d]	[2, 83, 240, 1466–1468]
			207.0 ± 6.0	[16, 274]
			209.2	[232]
			222.0 ± 8.4	[212]
Standard molar entropy[e] (at 298.15 K and 100 kPa)	S°_{298}	J mol^{-1} K^{-1}	33.05 ± 2.09	[244]
			33.15	[202]
			33.17[c]	[10, 11, 217]
			33.32	[1, 33, 208, 211, 222, 224, 241]
			33.80[f]	[249]
			35.56 ± 6.28	[32, 35, 42, 207, 223, 235]
			37.66	[247]
			38.91 ± 1.26	[248]
Molar enthalpy difference	$H_{298} - H_0$	kJ mol^{-1}	5.607[f]	[249]
			5.832[c]	[10, 11, 222]
			5.841	[202]
			5.862	[208, 211, 224]
Standard molar heat capacity[g] (at 298.15 K and 100 kPa)	$c^\circ_{p,298}$	J mol^{-1} K^{-1}	36.34	[207]
			37.72[c]	[4, 10, 11, 48]
			37.82	[202]
			37.90 ± 1.20	[1, 32, 33, 208, 211, 222, 224, 241, 244]
			38.07	[208, 209, 257, 266]
			40.68	[250]

(continued)

Table 5.4 (continued)

Characteristics	Symbol	Unit	Value	Reference
Specific heat capacity[h] (at 298.15 K)	c	J kg^{-1} K^{-1}	352.0	[45]
			367.1	[33, 798]
			368.0	[222, 325, 341–343]
			458.0	[31, 199]
Molar enthalpy (heat) of melting (at 298.15 K)	$\Delta H°_m$	kJ mol^{-1}	79.5	[224, 241, 1471]
Molar enthalpy (heat) of melting (at the melting point)	ΔH_m	kJ mol^{-1}	83.7	[33, 211, 235]
			113	[1507, 1508]
Specific enthalpy (heat) of melting (at the melting point)		kJ kg^{-1}	814	[33]
			850	[279]
			1100[i]	[1507, 1508]
Molar enthalpy (heat) of vaporization (at 298.15 K)[j]	$\Delta H°_v$	kJ mol^{-1}	608[k]	[74]
			805	[229]
Molar enthalpy (heat) of vaporization (dissociation)[l]	$\Delta \hat{H}_v$	kJ mol^{-1}	1515	[10]
			1520[m]	[219, 251]
Melting point	T_m	K (°C)	3400 ± 50 (3130 ± 50)	[215]
			3450 ± 50 (3175 ± 50)	[35, 246, 319]
			3670 (3400)	[704]
			3690 ± 20 (3420 ± 20)[n,o,p]	[4, 10, 11, 33, 45, 48, 72, 80, 211, 238, 263, 380, 1032]
			3700 (3430)[q]	[16, 267, 269, 279]
			3710 (3440)[r]	[3, 256]
			3720 ± 25 (3450 ± 25)[r,s]	[75, 271, 382, 635, 1009]
			3750 ± 35 (3475 ± 35)	[216]
			3755 (3480)	[227, 228]
			3775 (3500)	[377, 378]
			3800 ± 50 (3530 ± 50)[t]	[1, 30–32, 37, 42, 46, 47, 208, 213, 214, 223, 381, 659, 1412]
			3805 ± 125 (3532 ± 125)	[224, 226, 241, 285, 477]
			3810 ± 70 (3535 ± 70)	[149, 201, 218]
			3815 (3540)[u]	[15, 149, 1471]
			3830 (3560)	[63, 252, 259]
			4010 (3735)	[207]
			4015 (3740)	[270]

(continued)

Table 5.4 (continued)

Characteristics	Symbol	Unit	Value	Reference
Boiling point	T_b	K (°C)	5130 (4860)	[238]
			5370 (5100)	[30, 32, 37, 39, 42, 45, 199, 207, 1471]
			5920 (5650)	[31]

[a]For liquid state: $-\Delta H°_{298}$ = 118.8 kJ mol^{-1} [224], 110.0 kJ mol^{-1} [211]; in gaseous state: $-\Delta H°_{298}$ = 1484 kJ mol^{-1} [208]; enthalpy (heat) of complete dissociation (atomization) from solid state at 298,15 K ($-\Delta_{at}H°_{298}$, kJ mol^{-1}): 1523 ± 7 [1, 232], 1506 [32], 1512 [233], 1565 [234], 1520 ± 10 [1468] (for ZrC_{1-x} phases with different deviations from the stoichiometry: 1489 ± 9 (x = 0.04), 1437 ± 9 (x = 0.10), 1350 ± 8 (x = 0.20), 1263 ± 9 (x = 0.30), 1180 ± 10 (x = 0.40) [1]; 1509 (x = 0.04), 1359 (x = 0.20) [276])

[b]First principles calculated with generalized gradient approximation (GGA) using the Perdew-Burke-Ernzerhof (PBE) functional

[c]Carbon content – 49.0 at.%

[d]Extrapolated to the stoichiometric composition

[e]Molar entropy $S°_T$ (at 1200 K), J mol^{-1} K^{-1}, for ZrC_{1-x} phases with different deviations from the stoichiometry: 89.51 (x = 0), 88.54 (x = 0.05), 87.57 (x = 0.10), 86.60 (x = 0.15), 85.64 (x = 0.20), 84.67 (x = 0.25), 83.70 (x = 0.30), 82.73 (x = 0.35) and 81.76 (x = 0.40) [202]

[f]Calculated by means of the Lindemann equation using a method similar to that utilized for TiB_2

[g]Molar heat capacity $c_{p,T}$ (at 1200 K), J mol^{-1} K^{-1}, for ZrC_{1-x} phases with different deviations from the stoichiometry: 48.75 (x = 0), 47.53 (x = 0.05), 46.30 (x = 0.10), 45.08 (x = 0.15), 43.56 (x = 0.20), 42.63 (x = 0.25), 41.40 (x = 0.30), 40.18 (x = 0.35) and 38.96 (x = 0.40) [202]

[h]Specific heat capacity c, J kg^{-1} K^{-1}, at elevated and high temperatures for ZrC_{1-x} phases with different deviations from the stoichiometry: $ZrC_{\sim 1.0}$ – 630 (2230 °C) [74, 105]; $ZrC_{0.95}$ – 477 (1130 °C), 515 (1530 °C), 552 (1930 °C), 594 (2330 °C); $ZrC_{0.85}$ – 502 (1130 °C), 561 (1530 °C), 623 (1930 °C), 682 (2330 °C) [239]

[i]For $ZrC_{0.95}$ (temperatures of solidus – 3450 (3180) K (°C) and liquidus – 3850 (3580) K (°C))

[j]Enthalpy (heat) of sublimation of metallic Zr at 298,15 K, $\Delta_{sMe}H°_{298}$ = 599.3 kJ mol^{-1} [1469]

[k]Based on Knudsen mode measurements

[l]The sum of average partial enthalpies (heats) of vaporization of C_1 (gas) and Zr (gas) from $ZrC_{\sim 1.0}$ calculated in accordance to the second law of thermodynamics

[m]Based on Langmuir mode measurements

[n]Carbon content – 44.4 at.% [10]

[o]Carbon content – 49.2 at.% [11]

[p]Carbon content of ~46 at.% is corresponding to the maximal melting temperature; for 49.9 at.% of carbon composition: solidus and liquidus temperatures are ~3375 and ~3410 °C, respectively [72]

[q]Carbon content – 46 at.%

[r]Carbon content – 44–46 at.%

[s]Carbon content – 44.1 at.%

[t]Carbon content – 48.7 at.%

[u]Carbon content – 46.5 at.%

in the range of temperatures from 300 to 3700 K (up to melting point) [211]

$$c_p = 47.43 + \left(9.234 \times 10^{-3}\right)T - \left(1.092 \times 10^{6}\right)T^{-2}, \tag{5.8}$$

in the range of temperatures from 300 to 3800 K (up to melting point) [224]

$$c_p = 51.03 + \left(3.686 \times 10^{-3}\right)T - \left(1.990 \times 10^{-7}\right)T^2 + \left(2.893 \times 10^{-11}\right)T^3 - \\ - \left(1.304 \times 10^{6}\right)T^{-2}, \tag{5.9}$$

in the range of temperatures from 500 to 2870 K [245]

$$c_p = 73.21 - \left(3.591 \times 10^{-2}\right)T + \left(1.348 \times 10^{-5}\right)T^2, \tag{5.10}$$

in the range of temperatures from 1000 to 3000 K [202, 204]

$$c_p = 45.99 + \left(4.841 \times 10^{-3}\right)T + \left(8.368 \times 10^{5}\right)T^{-2}, \tag{5.11}$$

in the range of temperatures from 1400 to 2800 K [54]

$$c_p = 38.50 + \left(0.962 \times 10^{-2}\right)T, \tag{5.12}$$

in the range of temperatures from 1400 to 2800 K [54]

$$c_p = 80.64 \exp\left(-641.5/T\right), \tag{5.13}$$

in the range of temperatures from 1600 to 2300 K [202, 205]

$$c_p = 38.41 + \left(0.839 \times 10^{-2}\right)T. \tag{5.14}$$

in the range of temperatures from 3800 to 6000 K (in liquid state) [224]

$$c_p = 62.76 - \left(2.984 \times 10^{-10}\right)T + \left(5.524 \times 10^{-14}\right)T^2 - \left(3.538 \times 10^{-18}\right)T^3 - \\ - 0.8337T^{-2}, \tag{5.15}$$

For the molar heat capacity $c_p = f(T, \text{K})$, J mol^{-1} K^{-1}, of zirconium monocarbide phases with various deviations from the stoichiometry, the following relationships were recommended in literature:

for $ZrC_{0.58}$ (in the range of temperatures from 700 to 2500 K) [4, 210]

$$c_p = 28.7 + \left(1.06 \times 10^{-3}\right)T, \tag{5.16}$$

for $ZrC_{0.69}$ (in the range of temperatures from 1200 to 2500 K) [4, 202, 208, 1465]

$$c_p = 33.71 + \left(8.067 \times 10^{-3}\right)T + \left(31.90 \times 10^{12}\right)T^{-2}\exp(-31400/T), \tag{5.17}$$

for $ZrC_{0.76}$ (in the range of temperatures from 1200 to 2500 K) [4, 202, 208, 1465]

$$c_p = 32.48 + \left(9.824 \times 10^{-3}\right)T + \left(13.85 \times 10^{10}\right)T^{-2}\exp(-19960/T), \tag{5.18}$$

for $ZrC_{0.80}$ (in the range of temperatures from 700 to 2500 K) [4, 210]

$$c_p = 25.1 + \left(1.26 \times 10^{-3}\right)T, \tag{5.19}$$

for $ZrC_{0.96}$ (in the range of temperatures from 300 to 2500 K) [31, 96, 203, 207]

$$c_p = 36.44 + \left(9.540 \times 10^{-3}\right)T - \left(2.636 \times 10^{5}\right)T^{-2}, \tag{5.20}$$

for $ZrC_{0.96}$ (in the range of temperatures from 300 to 3200 K) [74, 84]

$$c_p = 56.86 - \left(10.90 \times 10^{-3}\right)T + \left(5.586 \times 10^{-6}\right)T^2 - \left(1.456 \times 10^{6}\right)T^{-2}, \tag{5.21}$$

for $ZrC_{0.96}$ (in the range of temperatures from 1300 to 2800 K) [94, 202]

$$c_p = 41.50 + \left(0.954 \times 10^{-3}\right)T, \tag{5.22}$$

for $ZrC_{0.96}N_{0.04}$ (in the range of temperatures from 300 to 1100 K) [278]

$$c_p = 48.18 + \left(5.35 \times 10^{-3}\right)T - \left(1.851 \times 10^{-6}\right)T^2 - \left(1.202 \times 10^{6}\right)T^{-2}, \tag{5.23}$$

for $ZrC_{0.97}$ (in the range of temperatures from 700 to 2500 K) [4, 210]

$$c_p = 24.5 + \left(0.80 \times 10^{-3}\right)T, \tag{5.24}$$

for $ZrC_{0.98}$ (in the range of temperatures from 1200 to 2500 K) [202, 203]

$$c_p = 25.18 + 0.183T, \tag{5.25}$$

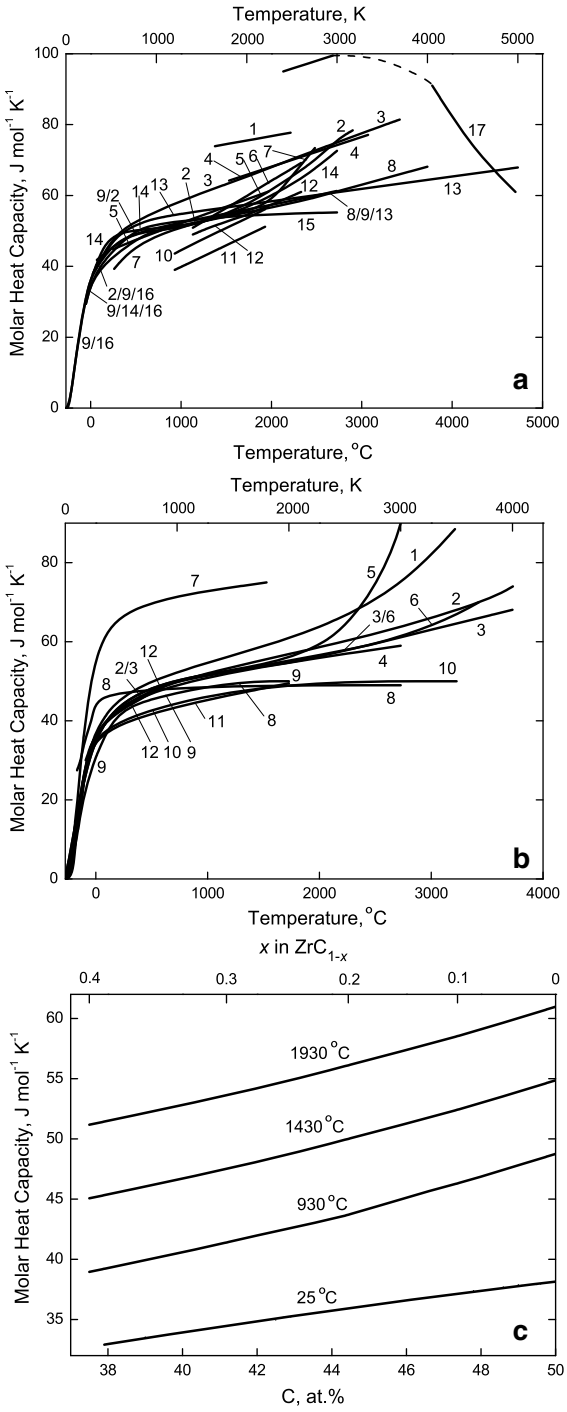

Fig. 5.5 (see caption on page 439)

◂**Fig. 5.5** Variations of molar heat capacity c of zirconium carbide ZrC_{1-x} phases with temperature (**a–b**) and carbon content within the homogeneity range at various temperatures (**c**): (**a**) – experimental data at constant pressure, c_p (1 – $ZrC_{0.94}$, contents: O – 1.21%, N – 1.03%, Hf – 0.78% [254]; 2 – $ZrC_{0.96}$, on the basis of several sources [84, 94, 255, 265, 278]; 3 – $ZrC_{\sim 1.0}$, recommended on the basis of several sources [211]; 4 – $ZrC_{\sim 1.0}$ [275]; 5 – plasma-sprayed $ZrC_{0.95}$, non-combined C – 0.67% [253]; 6 – $ZrC_{0.85}$ [239]; 7 – $ZrC_{0.92–0.94}$, contents: O < 0.1%, N – 0.5% [245]; 8 – $ZrC_{\sim 1.0}$, assessment based on thermodynamical calculations [16], experimentally confirmed [204, 205]; 9 – $ZrC_{\sim 1.0}$, 10 – $ZrC_{0.80}$ and 11 – $ZrC_{0.60}$, calculated on the basis of experimental data [202, 208, 257, 258]; 12 – $ZrC_{0.95}$ [239]; 13 – $ZrC_{\sim 1.0}$, recommended on the basis of several sources [224, 241]; 14 – $ZrC_{0.96}$, on the basis of several sources [10]; 15 – $ZrC_{\sim 1.0}$, recommended on the basis of several sources [272]; 16 – $ZrC_{0.95}$, contents: O – 0.005%, N – 0.067% [222] and 17 – $ZrC_{0.95}$, from experimental data on pulsed electrical heating with microsecond duration around solidus (3180 °C) and liquidus (3580 °C) temperatures with uncertainties ± 60 K for solid and ± 110 K for liquid states [1507, 1508]); (**b**) – theoretically calculated values for stoichiometric $ZrC_{\sim 1.0}$ at constant pressure, c_p (1 – Slater's approximation and 2 – Dugdale-MacDonald's approximation on the basis of Debye-Grüneisen model [280], 3 – thermodynamical assessment [16], 4 – on the basis of the projector augmented-wave (PAW) method within the generalized gradient approximation (GGA) [284], 5 – on the basis of two-stage upsampled thermodynamic integration using Langevin dynamics (TU-TILD) approach within GGA and 6 – same, within local density approximation (LDA) [698]) and at constant volume, c_v, on the basis of density functional theory (DFT) (7 – [69], 8 – [285], 9 – [284], 10 – [281], 11 – using GGA of the Perdew-Burke-Ernzerhof (PBE) scheme, lattice heat capacity [282] and 12 – [283]); (**c**) – experimental data at constant pressure, c_p (at 25 °C, based on several sources [84, 85, 222]; at 930, 1430 and 1930 °C, statistical-mathematical treatment of experimental temperature-concentration functions [202, 206, 208, 209, 257, 258]); when it is not indicated specially, data given are for quasi-stoichiometric composition

for $ZrC_{0.99}$ (in the range of temperatures from 1200 to 2500 K) [4, 202, 208, 1465]

$$c_p = 30.66 + \left(13.92 \times 10^{-3}\right)T \tag{5.26}$$

and for ZrC_{1-x} in general (in the range of temperatures from 1200 to 2200 K, for $0 < x \leq 0.31$) [202, 206]

$$c_p = (1 - 0.427x)\left[30.765 + \left(14.108 \times 10^{-3}\right)T\right], \tag{5.27}$$

or

$$c_p = 34.077 + \left(12.288 \times 10^{-3}\right)T - 24.483x, \tag{5.28}$$

where T is temperature, K, and x is the value of index in ZrC_{1-x} formula (*note*: Eq. (5.28) for $0 < x \leq 0.05$ can be applied in the extended interval at higher temperatures – up to 2500 K); for the specific heat capacity of zirconium monocarbide ZrC_{1-x} in the homogeneity range ($0 < x < 0.35$) at ~300 K the equation $c = f(T, \text{K})$, J kg^{-1} K^{-1}, is given by Katoh et al. [96] as

$$c = 373.2 - 355.6x + 686.0x^2 - 335.3x^3. \tag{5.29}$$

The variations of molar heat capacity c with temperature for zirconium monocarbide ZrC_{1-x} phases are demonstrated on the basis of several sources in Fig. 5.5. The thermal properties of ZrC_{1-x} are very sensitive to the deviation from the stoichiometry: for the standard heat of formation ΔH°_{298}, kJ mol^{-1}, molar enthalpy difference $H^\circ_T - H^\circ_{298}$, J mol^{-1}, and molar entropy S°_T, J mol^{-1} K^{-1}, the following concentration and temperature-concentration dependencies within the homogeneity range of ZrC_{1-x} were calculated on the basis of experimental data [2, 202, 206, 208, 209, 240, 257, 258, 286, 292, 1466–1468]:

$$\Delta H^\circ_{298} = -206.7 + 154.04x, \tag{5.30}$$

$$\Delta H^\circ_{298} = 27.97 - 347.4(1-x) + 112.8(1-x)^2 \pm 10.0, \tag{5.31}$$

$$H^\circ_T - H^\circ_{298} = (1 - 0.427x)\left[30.765T + \left(7.054 \times 10^{-3}\right)T^2 - 3230\right], \tag{5.32}$$

$$H^\circ_T - H^\circ_{298} = 12,713x - 24.488xT + 34.077T + \left(6.114 \times 10^{-3}\right)T^2 - 6198 \tag{5.33}$$

$$S^\circ_T = 74.48 \lg T + \left(12.228 \times 10^{-3}\right)T + 151.21x - 55.40x \lg T - 154.50, \tag{5.34}$$

where T is temperature, K, and x is the value of index in ZrC_{1-x} formula (in general Eqs. (5.32–5.34) were recommended for the range of temperatures from 1200 to 2200 K, but for carbon-rich compositions ($0 \le x \le 0.05$) the interval of reliable usage expands up to 2500 K). The thermodynamic functions of stoichiometric zirconium carbide are tabulated by Turchanin et al. [202, 208, 257] in the range of 0–3000 K, by Schick [211] and Chase [224] in the range of 0–6000 K, and by Barin [241] in the range of 298.15–4000 K, the thermodynamic functions of $ZrC_{0.96}$ in the range of 298.15–3000 K – by Storms [10] and Toth [11].

During the vaporization processes from the surface of zirconium monocarbide at high and ultra-high temperatures in vacuum, the composition of the carbide phase (C/Zr ratio) can change noticeably. The following equation was recommended for zirconium (P_{Zr}, Pa) partial pressure over near-stoichiometric monocarbide phase, which is in equilibrium with carbon (ZrC_{1-x} + C), by Storms [10] for 2520–2820 K (2245–2545 °C):

$$\lg P_{Zr} = -\left(4.161 \times 10^4\right)/T + 12.592, \tag{5.35}$$

the estimation of zirconium (P_{Zr}, Pa) and carbon (P_C, Pa) partial pressures in the gaseous phase in the Zr–C system carried out by Kulikov [238] led to such relationships as

for the conditions of congruent vaporization of quasi-stoichiometric ZrC_{1-x}:

at 1000–2125 K (730–1850 °C)

$$\lg P_{Zr} = -\left(3.948 \times 10^4\right)/T + 12.888, \tag{5.36}$$

at 2125–3690 K (1850–3420 °C)

$$\lg P_{Zr} = -\left(3.933 \times 10^4\right)/T + 12.819, \tag{5.37}$$

at 3690–5000 K (3420–4730 °C)

$$\lg P_{Zr} = -\left(3.379 \times 10^4\right)/T + 11.319, \tag{5.38}$$

for the monoatomic gases of Zr and C over ZrC_{1-x} in the monocarbide homogeneity range:

at 2000 K (1730 °C)

$$\lg P_{Zr} = -0.935 - 7.420(1 - x)^2, \tag{5.39}$$

$$\lg P_C = -12.792 + 7.420(1 - x)^2, \tag{5.40}$$

at 3000 K (2730 °C)

$$\lg P_{Zr} = 3.236 - 5.069(1 - x)^2, \tag{5.41}$$

$$\lg P_C = -3.971 + 5.069(1 - x)^2, \tag{5.42}$$

at 5000 K (4730 °C)

$$\lg P_{Zr} = 5.467 - 1.904(1 - x)^2, \tag{5.43}$$

$$\lg P_C = 3.473 + 1.904(1 - x)^2, \tag{5.44}$$

for the conditions of ZrC_{1-x}–C phases equilibrium:

at 1000–2125 K (730–1850 °C)

$$\lg P_{Zr} = -\left(4.143 \times 10^4\right)/T + 12.534, \tag{5.45}$$

at 2125–3690 K (1850–3420 °C)

$$\lg P_{Zr} = -\left(4.057 \times 10^4\right)/T + 12.122, \tag{5.46}$$

at 3690–5000 K (3420–4730 °C)

$$\lg P_{Zr} = -\left(3.430 \times 10^4\right)/T + 10.423, \tag{5.47}$$

where T is temperature, K and x is the value of index in ZrC_{1-x} (*see* Tables 5.5 and 5.6). Within the homogeneity range of ZrC_{1-x} the vapour pressure of metal over the carbide surface increases rapidly, e.g. from $\sim 4.1 \times 10^{-8}$ Pa for $ZrC_{\sim 1.0}$ + C to $\sim 2.0 \times 10^{-3}$ Pa for $ZrC_{0.58}$ + Zr at 2100 K (1830 °C), or from $\sim 9.1 \times 10^{-2}$ Pa up to ~ 20 Pa for the similar compositions at 3000 K (2730 °C), while the average partial enthalpy (heat) of vaporization of zirconium decreases from ~ 820–840 kJ mol^{-1} for $ZrC_{\sim 1.0}$ to ~ 570–590 kJ mol^{-1} for $ZrC_{0.58}$ (the sharpest drop in this value is observed for compositions within $0 < x < 0.1$ interval) [10, 26, 27, 84, 294, 1377]. The partial pressures of zirconium and carbon over the monocarbide phase as functions of composition at the fixed temperature 3000 K (2730 °C) are shown in Fig. 5.6 in comparison with the equilibrium pressure of carbon over pure graphite surface. The vapour phase above carbon-saturated zirconium monocarbide in equilibrium with graphite contains atoms of zirconium and molecules of carbon [293], plus molecules of ZrC_2, observed at 2660 K (2390 °C) by Starostina et al. [289]. The equilibrium pressure should be lower than the pressure observed upon vaporization from an open surface into a vacuum (Langmuir mode). According to Nikolskaya et al. [290, 291] the vapour pressures of zirconium (P_{Zr}, Pa) and carbon (P_C, Pa) in Langmuir vaporization of zirconium monocarbide in the temperature range of 2300–3100 K (2030–2830 °C) can be desribed by the equations

$$\lg P_{Zr} = 11.26 - \left(3.147 \times 10^4\right)/T - (7.2 \times 10^{-5})T + \lg x, \tag{5.48}$$

$$\lg P_C = 14.76 - \left(4.759 \times 10^4\right)/T + (1.2 \times 10^{-5})T + \lg[(1 - x)/x], \tag{5.49}$$

where x is the value of index in ZrC_{1-x} formula and T is temperature, K. For the general evaporation rate G, g cm^{-2} s^{-1}, as a function of temperature T, K, the following equations were proposed by Fesenko and Bolgar [26, 27, 208, 288] with the measurements from an open surface into a vacuum for $ZrC_{0.95}$ in the range of temperatures from 2770 K (2500 °C) to 3170 K (2900 °C):

$$\lg G = (8.9 \pm 0.2) - [(4.11 \pm 0.04) \times 10^4]/T \tag{5.50}$$

and by Vlasov et al. [295] for $ZrC_{0.97}$ (lost mainly carbon during the initial heating) in the range of temperatures from 2500 K (2230 °C) to 3270 K (3000 °C)

Table 5.5 Parameters of the gaseous phase in the Zr−C system in the conditions of Zr−ZrC_{1-x} phases equilibrium, congruent dissociation of quasi-stoichiometric ZrC_{1-x} and ZrC_{1-x}−C phases equilibrium calculated on the basis of thermodynamic data [238]

Parameters	Temperature, K (°C)								
	1000 (730)	1500 (1230)	2125 (1850)	2500 (2230)	3000 (2730)	3690 (3420)	4000 (3730)	4645 (4370)	5000 (4730)
					Zr−ZrC_{1-x} phases equilibrium				
lgP_{Zr}, Pa	−18.885	−8.582	−2.504	−0.396	1.580	3.416	4.021	5.025	–
lgP^{a}_{Σ}, Pa	−18.885	−8.582	−2.504	−0.396	1.580	3.417	4.024	5.031	–
lg(C/Zr)[b]	−15.412	−9.862	−6.245	−4.942	−3.655	−2.469	−2.180	−1.857	–
Contents, vol.%:									
C	–	–	–	–	0.02	0.21	0.39	0.77	–
C_2	–	–	–	–	–	–	0.005	0.01	–
Zr	100.0	100.0	100.0	100.0	99.97	99.66	99.34	98.63	–
ZrC[c]	–	–	–	–	0.007	0.13	0.26	0.59	–
					Congruent dissociation of quasi-stoichiometric ZrC_{1-x}				
lgP_{Zr}, Pa	−26.591	−13.416	−5.691	−2.908	−0.282	2.168	2.918	4.052	4.560
lgP^{a}_{Σ}, Pa	−26.291	−13.115	−5.393	−2.617	−0.005	2.427	3.178	4.322	4.833
Contents, vol.%:									
C	50.00	49.94	49.18	47.20	42.76	37.18	34.83	35.28	35.21
C_2	–	0.01	0.31	0.93	2.30	4.28	5.00	5.46	5.72
C_3	–	0.003	0.17	0.67	1.84	3.21	3.37	2.47	2.20
Zr	50.00	50.05	50.33	51.16	52.84	55.09	54.95	53.76	53.30
ZrC[c]	–	–	0.01	0.04	0.26	1.25	1.84	3.01	3.55

(continued)

Table 5.5 (continued)

Parameters	Temperature, K (°C)								
	1000 (730)	1500 (1230)	2125 (1850)	2500 (2230)	3000 (2730)	3690 (3420)	4000 (3730)	4645 (4370)	5000 (4730)
				ZrC_{1-x}–C phases equilibrium					
$\lg P_{Zr}$, Pa	−28.896	−15.067	−6.972	−4.068	−1.357	1.135	1.857	3.038	3.563
$\lg P_{\Sigma}^{a}$, Pa	−24.286	−11.719	−4.044	−1.154	1.596	4.156	5.035	6.012	6.394
lg(C/Zr)[b]	4.610	3.426	3.250	3.161	3.357	3.388	3.521	3.252	3.076
Contents, vol.%:									
C	99.88	89.72	42.12	23.51	12.75	7.28	5.58	7.76	9.61
C_2	0.01	0.82	5.08	6.66	8.15	9.28	9.21	12.97	15.52
C_3	0.11	9.41	52.66	69.23	77.34	75.05	71.36	63.36	59.28
C_4	–	–	0.01	0.04	0.26	1.25	2.04	3.16	3.62
C_5	–	–	0.02	0.20	1.38	7.02	11.71	12.58	11.72
Zr	0.003	0.05	0.12	0.12	0.11	0.10	0.07	0.11	0.15
ZrC[c]	–	–	–	0.002	0.007	0.02	0.03	0.06	0.10

[a]Total gas pressure
[b]Logarithm of atomic ratio in gaseous phase
[c]Gaseous molecule

Table 5.6 Calculated parameters of the gaseous phase over zirconium monocarbide ZrC_{1-x} in its homogeneity range on the basis of thermodynamic data [238]

Parameters	Phase composition, x													
	0.01	0.04	0.10	0.11	0.125	0.15	0.17	0.19	0.20	0.25	0.30	0.35	0.40	0.42
						2000 K (1730 °C)								
$\lg P_{Zr}$, Pa	−8.207	–	–	−6.839	−6.613	−6.296	–	–	–	−5.109	–	−4.070	–	−3.431
$\lg P^{a}_{\Sigma}$, Pa	−5.223	–	–	−6.561	−6.494	−6.265	–	–	–	−5.109	–	−4.070	–	−3.431
Contents, vol.%:														
C	50.52	–	–	47.07	23.97	6.83	–	–	–	0.03	–	–	–	–
C_2	4.04	–	–	0.16	0.05	0.007	–	–	–	–	–	–	–	–
C_3	45.34	–	–	0.08	0.01	0.001	–	–	–	–	–	–	–	–
Zr	0.10	–	–	52.69	75.97	93.16	–	–	–	99.97	–	100.0	–	100.0
						3000 K (2730 °C)								
$\lg P_{Zr}$, Pa	–	−1.436	−0.870	–	–	–	−0.291	−0.071	−0.008	–	0.752	–	1.411	–
$\lg P^{a}_{\Sigma}$, Pa	–	1.577	0.370	–	–	–	−0.046	0.021	0.069	–	0.755	–	1.411	–
Contents, vol.%:														
C	–	12.99	58.18	–	–	–	40.01	21.59	15.99	–	0.57	–	0.03	–
C_2	–	8.29	10.08	–	–	–	1.83	0.62	0.15	–	0.002	–	–	–
C_3	–	78.63	25.99	–	–	–	1.24	0.27	0.13	–	–	–	–	–
Zr	–	0.09	5.75	–	–	–	56.92	77.52	87.73	–	99.43	–	99.97	–

[a]Total gas pressure

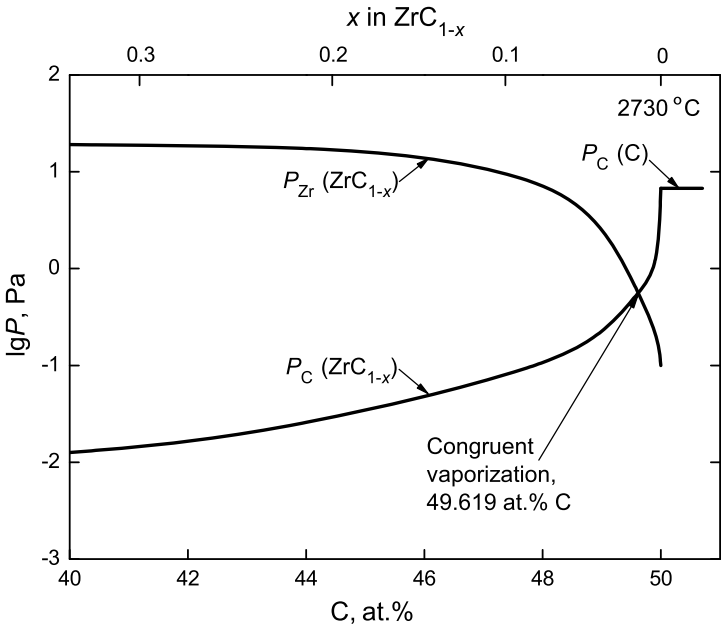

Fig. 5.6 Partial pressures of zirconium P_{Zr} (ZrC_{1-x}) and carbon P_C (ZrC_{1-x}) over zirconium carbide ZrC_{1-x} phase, and vapour pressure of carbon over pure graphite P_C (C) as functions of carbide composition at 3000 K (2730 °C) calculated by Kaufman [11, 33, 231, 287]; a congruently vaporizing composition is marked

$$\lg G = (3.2 \pm 0.5) - [(2.26 \pm 0.24) \times 10^4]/T \tag{5.51}$$

and for $ZrC_{0.88}$ (demonstrated the constancy of the composition irrespectively of the duration of heating) at 2600–3030 K (2330–2760 °C)

$$\lg G = (8.1 \pm 1.2) - [(3.92 \pm 0.35) \times 10^4]/T, \tag{5.52}$$

where T is temperature, K. Through a large part of the homogeneity range of ZrC_{1-x} the vapour phase should receive metallic atoms mainly, although in a certain area of compositions congruent vaporization may occur [26, 27, 296]. According to Kaufman and Sarney [287, 294], Coffman et al. [229] and Fesenko and Bolgar et al. [288, 1474], the congruently vaporizing composition of zirconium monocarbide phase at temperatures above 2500 K (2230 °C) corresponds to $ZrC_{\sim 1.0}$ ($x \approx 0$). However, at temperatures about 3000 K (2730 °C) Storms [10] showed that this carbide composition is located at around of $0.13 \leq x \leq 0.18$; the similar data, which could be described by the following equation for the congruently vaporizing composition of ZrC_{1-x}:

$$\lg x = -0.1696 - \left(1.633 \times 10^3\right)/T, \tag{5.53}$$

where x is the value of index in ZrC_{1-x} formula and T is temperature, K, were presented by Nikolskaya et al. [290, 291] for the temperature range from 2270 K (2000 °C) to 3070 K (2800 °C), where the composition vaporizing congruently shifts from $ZrC_{0.87}$ to $ZrC_{0.80}$ with temperature growth, while the rate of congruent vaporization V, g cm^{-2} s^{-1}, was estimated by them in the temperature interval of 2390–3100 K (2120–2830 °C) as

$$\lg V = 8.555 - \left(3.813 \times 10^4\right)/T, \tag{5.54}$$

where T is temperature, K. Also it should be mentioned that Andrievskii et al. [231, 298] demonstrated that the value of congruently vaporizing composition of ZrC_{1-x} phase at temperatures about 2300 K (2030 °C) ranges within the limited interval of $0.08 \leq x \leq 0.09$ and depends very little on temperature; while Zotov and Kotelnikov [1473], who studied volatilization of various compositions of zirconium monocarbide from $ZrC_{0.72}$ to $ZrC_{0.98}$ in high vacuum ($3 \times 10^{-2} - 5 \times 10^{-3}$ Pa) at 2770–3100 K (2500–2830 °C), revealed that the interval of congruently vaporizing compositions in above mentioned conditions is around $0.12 \leq x \leq 0.14$ and a decrease in carbon content in ZrC_{1-x} increases vaporization rate of the materials noticeably [320].

The values of some general thermodynamic properties, vapour pressures and mass/linear vaporization rates for zirconium monocarbide are given in Addendum in comparison with other ultra-high temperature materials in the wide ranges of temperatures.

At room temperature the thermal conductivity of near-stoichiometric zirconium monocarbide ZrC_{1-x} is around 15–40 W m^{-1} K^{-1} (with thermal diffusivity of about 0.07–0.15 cm^2 s^{-1}) [1, 26, 27, 30, 32, 35, 37, 39, 45, 46, 53, 85, 199, 213, 278, 299–301, 306, 308, 319, 320, 325, 328, 331–337, 341, 342, 377, 378, 380, 721]; it is affected by the incoherent boundaries and porosity in carbide materials noticeably [96], e.g. the thermal conductivity of ZrC_{1-x} fibrous materials with 70–75% porosity at 300–900 °C is 0.5–2.0 W m^{-1} K^{-1} [336–339]. Marmer et al. [213] revealed for ZrC_{1-x} materials in the range of 1200–2700 °C the increase of thermal conductivity from 9–10 to 19–38 W m^{-1} K^{-1} with 12–20% increase of porosity, it should be mentioned especially that the same researchers also found out the similar effect of porosity on NbC_{1-x} and TiC_{1-x} materials. Within the homogeneity range, as a consequence of conduction electrons scattering on the carbon sublattice vacancies and thermal lattice vibrations, the thermal conductivity of ZrC_{1-x} declines with increasing carbon deficit (value of x) in the phase [33, 344] that is demonstrated perfectly in Fig. 5.7; while Wagner [345] proposed that the electronic contribution to the thermal conductivity at room temperature, which is ~37% for a material with composition in the range of $0.15 \leq x \leq 0.35$, diminishes as x approaches zero. Storms and Wagner [85] gave an empirical equation ($0 < x \leq 0.4$)

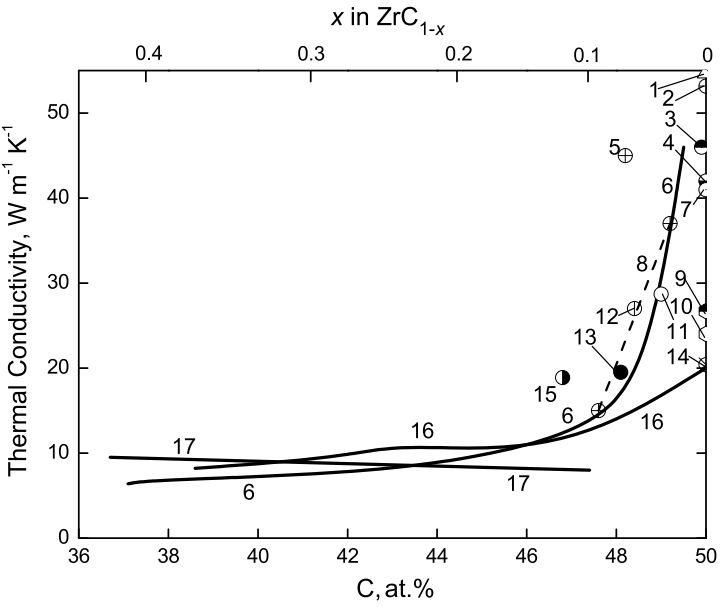

Fig. 5.7 Thermal conductivity λ of zirconium monocarbide ZrC_{1-x} materials at room temperature as a function of carbide composition within the homogeneity range: 1 – [30, 213]; 2 – at 300 °C [45, 306]; 3 – pyrolytic [301]; 4 – at 0 °C [53]; 5 – sintered, porosity 6–8%, contents: non-combined C – 0.5%, W – 0.2% [411]; 6 – hot-pressed, content O – 0.015–0.090%, corrected to porosity [85, 325]; 7 – at 50 °C, hot-pressed rods, corrected to porosity [308]; 8 – sintered, porosity 13–19% [297]; 9 – hot-pressed, porosity 1% [341, 342]; 10 – extrapolated value [328]; 11 – pyrolitic, content N – 0.03–0.08% [300]; 12 – hot-pressed, corrected to porosity [326]; 13 – hot-pressed, content O – 0.6% [90]; 14 – at 0–500 °C [35, 37, 46]; 15 – single crystal, contents: O < 0.01%, N < 0.01% [302, 303]; 16 – sintered in vacuum, corrected to porosity [1, 26, 27, 32, 299]; 17 – sintered in vacuum, content N – 0.05% [137]

$$\lambda = 1.05 \times 10^3 \left[3.82 \times 10^{-3} + (55 + 950x)^{-1}\right] + 0.07x^{-2}, \tag{5.55}$$

where x is the value of index in ZrC_{1-x} formula, to represent the relationship between the thermal conductivity λ and composition within the homogeneity range of monocarbide. However, the effect of an increase in the deviation from the stoichiometry was proposed by Avgustinik et al. [137] to reduce the connectivity of the monocarbide lattice while introducing vacancies and increasing the concentration of non-localized electrons, as they found out the decline of thermal conductivity from ~9 W m^{-1} K^{-1} for $ZrC_{0.60}$ to ~6 W m^{-1} K^{-1} for $ZrC_{0.90}$ employing the high-purity zirconium monocarbide materials. The variations of thermal conductivity with temperature for near-stoichiometric ZrC_{1-x} materials on

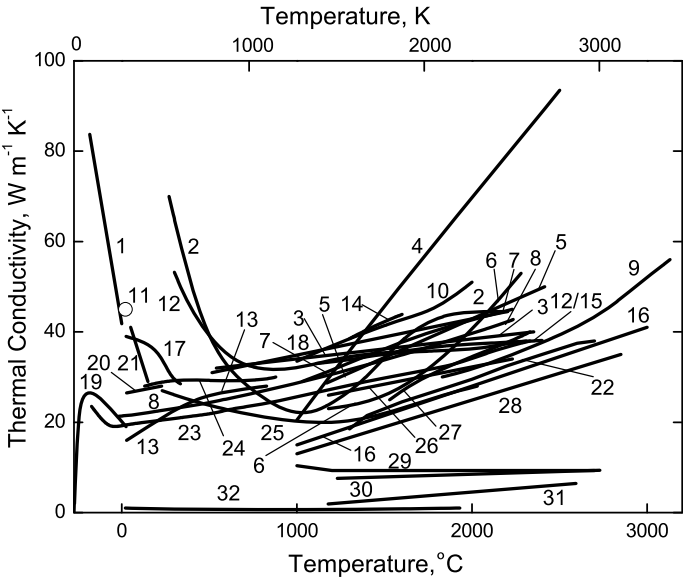

Fig. 5.8 Variation of thermal conductivity λ with temperature for near-stoichiometric zirconium monocarbide ZrC_{1-x} materials on the basis of several sources: 1 – $ZrC_{\sim 1.0}$, porosity 2% [321]; 2 – $ZrC_{0.92}$, sintered and annealed at 2200–2500 °C, porosity 7.5% and 3 – same, without annealing [245, 319]; 4 – $ZrC_{\sim 1.0}$, sintered and annealed in vacuum, porosity 10% [117, 329]; 5 – $ZrC_{0.96}$, porosity 5% [330]; 6 – $ZrC_{\sim 1.0}$, porosity 6% [337]; 7 – $ZrC_{0.97}$, sintered and annealed in vacuum, corrected to porosity, content N – 0.04% [210, 305]; 8 – [304, 320]; 9 – $ZrC_{0.96}$, porosity 12–16% [311]; 10 – pyrolytic $ZrC_{0.99}$ [314]; 11 – sintered $ZrC_{0.93}$, porosity 6–8%, contents: non-combined C – 0.5%, W – 0.2% [411]; 12 – [45, 306]; 13 – hot-pressed $ZrC_{0.96}N_{0.04}$, porosity 1% [278]; 14 – hot-pressed $ZrC_{0.93–0.99}$, content non-combined C – 0–0.8%, radial diffusivity method [322, 323]; 15 – $ZrC_{0.95}$ [239]; 16 – recommended on the basis of several sources [33]; 17 – hot-pressed $ZrC_{\sim 1.0}$ spheres, corrected to porosity [308]; 18 – hot-pressed $ZrC_{0.93–0.99}$, content non-combined C – 0–0.8, steady-state method [307]; 19 – single crystal $ZrC_{0.88}$, contents: O < 0.01%, N < 0.01% [302, 303]; 20 – [327]; 21 – hot-pressed $ZrC_{\sim 1.0}$ rods, corrected to porosity [308]; 22 – hot-pressed $ZrC_{0.93}$, porosity 19–24% [310, 324]; 23 – hot-pressed $ZrC_{0.92}$, porosity 5%, content O – 0.6% [90]; 24 – $ZrC_{0.79}$, porosity 5–12%, content non-combined C – 1.1% [312]; 25 – sintered $ZrC_{0.98}$, porosity 3–7% [313]; 26 – hot-pressed (in vacuum) $ZrC_{\sim 1.0}$, corrected to porosity and thermal expansion, contents: non-combined C – 0.18–0.30%, O – 0.001–0.09% [77]; 27 – pyrolytic $ZrC_{0.99}$ [309, 318]; 28 – hot-pressed $ZrC_{\sim 1.0}$ [26, 27]; 29 – sintered $ZrC_{0.98}$, porosity 5–7% [213]; 30 – sintered $ZrC_{0.98}$, corrected to porosity [315–317]; 31 – $ZrC_{\sim 1.0}$, porosity 63% [330]; 32 – unwoven felt $ZrC_{\sim 1.0}$, mean filament diameter 16 μm, porosity ~80% [338] (when it is not indicated specially, data are given for near-stoichiometric compositions)

the basis of several sources are shown in Fig. 5.8. The thermal conductivity λ of hot-pressed (in vacuum) $ZrC_{0.97}$ (contents: non-combined C – 0.3%, O – 0.09%) in the range of temperatures from 1230 to 2130 °C Grossman [77] described by the following equation (with accuracy ±10%):

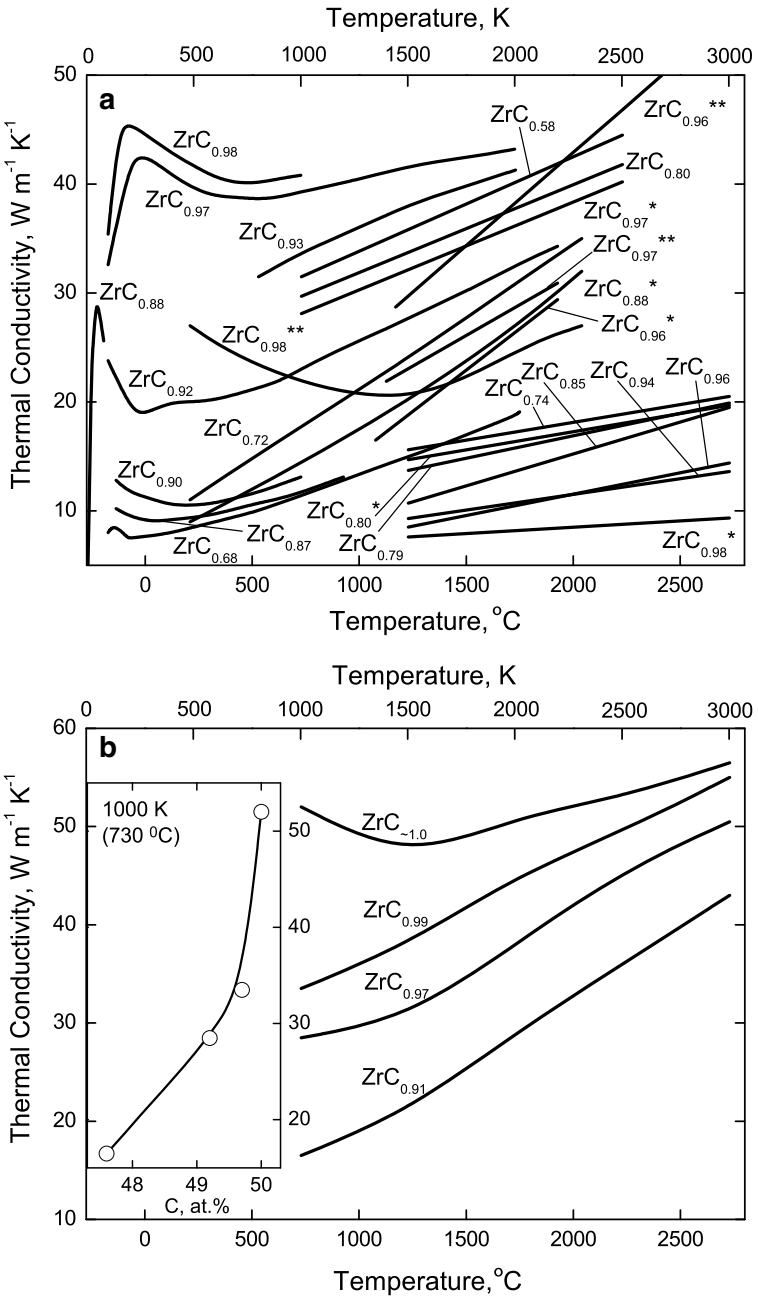

Fig. 5.9 (see caption on page 451)

◂**Fig. 5.9** Thermal conductivity λ of ZrC_{1-x} as a function of temperature for the phases having different deviations from the stoichiometry: (**a**) – experimental data (hot-pressed at 2730 °C, data corrected to 100% dense materials, content O – 0.015–0.1%, x = 0.02, 0.03, 0.10, 0.13, 0.32 [325]; sintered in vacuum, corrected to porosity, x = 0.03*, 0.20, 0.42 [210, 305, 409]; high purity materials sintered in vacuum at 2230 °C, corrected to porosity, content O + N – 0.1–0.3%, x = 0.02*, 0.04, 0.06, 0.15, 0.20*, 0.26 [61, 315–317]; hot-pressed, corrected to porosity, x = 0.07 [326]; single crystals, contents O, N < 0.01%, x = 0.12 [302, 303]; hot-pressed, porosity 5%, content O – 0.6%, x = 0.08 [90]; hot-pressed in vacuum, porosity 8.5%, contents: non-combined C – 0.3%, O – 0.09%, x = 0.03** [77]; hot-pressed, porosity 20%, x = 0.04* [310]; porosity 5%, x = 0.04** [330]; x = 0.02**, 0.12*, 0.28 [313]) and (**b**) – values calculated by empirical potential molecular dynamics (vibrational part) and density functional theory (electronic part) for x = 0, 0.01, 0.03, 0.09 (*Inset* – calculated relationship between thermal conductivity and carbon content in carbide for $0 \le x \le 0.09$ at 730 °C) [340]

$$\lambda = 16.7 + (7.11 \times 10^{-3})(t + 273), \tag{5.56}$$

and for the experimental data fit of thermal conductivity λ of $ZrC_{0.96}N_{0.04}$ phase Lengauer et al. [278] obtained the following polynomial expression for the range of 300–1100 K (25–830 °C):

$$\lambda = 22.11 + (7.224 \times 10^{-3})(t + 273) - (0.804 \times 10^{-6})(t + 273)^2 - \\ - (6.600 \times 10^5)/(t + 273)^2, \tag{5.57}$$

where t is temperature, °C. The experimentally measured and theoretically calculated thermal conductivities as functions of temperature for the various ZrC_{1-x} phases having different deviations from the stoichiometry are presented in Fig. 5.9.

At room temperature the mean coefficient of linear thermal expansion of near-stoichiometric zirconium monocarbide ZrC_{1-x} is $(4.0–7.4) \times 10^{-6}$ K^{-1} [47, 60, 63, 96, 199, 346, 357, 380, 452, 797, 887, 1480]. Usually, as it was confirmed by experiments, the porosity of materials has no significant effect on this property [345, 374, 384]. The experimental data collected from the various measurements of thermal expansion of zirconium monocarbide are listed in Table 5.7. The approximation functions for the temperature dependence of relative thermal linear expansion $\Delta l/l_0 = f(T, \text{K})$, %, of near-stoichiometric ZrC_{1-x} was recommended by Touloukian et al. [357] on the basis of experimental data for pure samples from several sources (accuracy within ±5% at $T < 2200$ K and within ±10% at $T > 2200$ K):

for 293–1500 K (20–1230 °C):

$$\Delta l/l_0 = -0.080 + (1.280 \times 10^{-4})T + (5.372 \times 10^{-7})T^2 - (1.392 \times 10^{-10})T^3, \tag{5.58}$$

Table 5.7 Average coefficients of linear thermal expansion α_m of zirconium monocarbide ZrC_{1-x} in various temperature ranges

Temperature range, °C	α_m, 10^{-6} K^{-1}	Reference
(–190)–(+20)	1.9 ± 0.8[a]	[47, 142]
	3.5 ± 0.4[b]	[47, 142]
	3.7 ± 0.4[c]	[47, 142]
(–170)–(+100)	4.51[d]	[134]
(–150)–(+30)	6.22[e]	[111]
(–20)–(+80)	6.2	[199]
(–20)–(+400)	8.1	[199]
20–100	4.57[f]	[357]
	5.45[d]	[134]
20–200	4.83[f]	[357]
	5.73[g]	[308]
	6.0	[45]
	7.2	[49]
20–300	5.17[f]	[357]
30–330	7.0[h]	[277]
20–400	5.47[f]	[357]
	6.0	[45]
	6.55[g]	[308]
20–500	5.75[f]	[357]
	6.73 ± 0.32[i]	[100]
	6.74	[246]
20–550	6.7–6.8[j]	[213, 320, 349, 350]
20–600	6.02[f]	[357]
	6.66	[45]
	6.80[i]	[100]
	6.94[g]	[308]
20–700	6.25[f]	[357]
	6.86 ± 0.22[i]	[100]
20–730	6.35	[242, 243]
20–800	6.47[f]	[357]
	6.7[k]	[37]
	6.75	[45]
	6.9[l]	[359]
	7.16[g]	[308]
20–850	6.93[i]	[100]
20–900	6.65[f]	[357]
	6.89[m]	[1350]
	7.42[n]	[1350]
	7.68[o]	[1350]

(continued)

Table 5.7 (continued)

Temperature range, °C	α_m, 10^{-6} K^{-1}	Reference
20–950	6.60[p]	[133]
	7.50[q]	[361]
20–1000	6.3[r]	[363]
	6.57 ± 0.08[s]	[57]
	6.72[t]	[60, 346, 635]
	6.80[f]	[357]
	6.87[u]	[130]
	6.9[v]	[356]
	6.91[w]	[348]
	6.99 ± 0.16[i]	[100]
	7.00	[45]
	7.01[s]	[1, 32, 364]
	7.05[x]	[384]
	7.14[y]	[373]
	7.33[g]	[308]
	7.79[z]	[372]
	8.73[a1]	[4, 91, 132]
	9.66[u]	[4, 91, 132]
20–1100	6.73	[30, 31, 378]
	6.93[f]	[357]
20–1130	7.00	[242, 243]
20–1200	6.53[a2]	[352, 365]
	7.00	[45]
	7.06[f]	[357]
	7.45[g]	[308]
	7.50[a3]	[51, 59, 469]
20–1300	6.91[a4]	[362]
	7.17[f]	[357]
	7.20[a5]	[347, 358]
20–1350	7.49[g]	[308, 369]
20–1400	6.70	[351, 355]
	6.77[a2]	[367]
	7.15	[45]
	7.29[f]	[357]
20–1500	7.40[d]	[357]
	7.43[a5]	[347, 358]
	9.15[w]	[348]
800–1500	8.43[a6]	[371]
20–1600	7.12	[45]
	7.51[f]	[357]
20–1700	7.16	[74, 81]
	7.60[f]	[357]
	7.70[a5]	[347, 358]

(continued)

Table 5.7 (continued)

Temperature range, °C	α_m, 10^{-6} K^{-1}	Reference
900–1700	7.4 ± 0.2[a7]	[81]
20–1730	7.0	[33]
20–1800	7.22	[45]
	7.70[f]	[357]
20–1900	7.78[f]	[357]
	8.10[a5]	[347, 358]
20–2050	7.04[a8]	[366]
20–2000	6.90	[45]
	7.56[t]	[60, 346]
	7.7[a9]	[360, 383]
	7.87[f]	[357]
	8.0[l]	[359]
	9.05[b1]	[368]
	9.77[w]	[348]
20–2100	4.63[b2]	[245]
	7.94[f]	[357]
1000–2100	9.13[b3]	[320, 354]
20–2130	7.98	[242, 243]
20–2200	7.45	[45]
	8.02[f]	[357]
20–2300	8.03[b4]	[370]
	8.10[f]	[357]
	8.77[a2]	[347, 358]
1000–2300	11.8[b5]	[319, 354]
20–2400	7.50	[45]
	8.17[f]	[357]
	8.65[v]	[356]
	10.34[w]	[348, 369]
20–2500	8.23[f]	[357]
20–2600	7.54	[45]
	8.30[f]	[357]
	9.30[a5]	[347]
20–2700	7.56 ± 0.20[b6]	[95, 320, 353]
	8.37[f]	[357]
	9.85[b7]	[326]
20–2900	9.61[a4]	[347]

[a]Powdered $ZrC_{0.80}$, measured by low-temperature X-ray diffraction method
[b]Powdered $ZrC_{0.97}$, measured by low-temperature X-ray diffraction method
[c]Powdered $ZrC_{0.95}$, measured by low-temperature X-ray diffraction method
[d]Sintered (in high vacuum) $ZrC_{0.97}$ (10–20% porosity), measured by dilatometric method
[e]Powdered $ZrC_{0.99}$, measured by low-temperature X-ray diffraction method
[f]Calculated on the basis of approximation function
[g]Hot-pressed $ZrC_{\sim 1.0}$ (9% porosity), measured (perpendicular to the direction of pressing) by dilatometric method in inert atmosphere
[h]Sintered (in vacuum) $ZrC_{0.98}$ (contents: non-combined C – 0.15–0.20%, O – 0.10–0.18%), measured by dilatometric method
[i]Powdered $ZrC_{0.96}$ (content non-combined C – 0.5%), measured by high-temperature X-ray diffraction method

(continued)

Table 5.7 (continued)

[j]Hot-pressed $ZrC_{0.83–0.85}$ (content non-combined C – 4.0%), measured using a dilatation interferometer
[k]Hot-pressed materials
[l]Pyrolytic $ZrC_{0.99}$
[m]Powdered $ZrC_{0.98}$, measured by high-temperature X-ray diffraction method
[n]Powdered $ZrC_{0.82}$, measured by high-temperature X-ray diffraction method
[o]Powdered $ZrC_{0.74}$, measured by high-temperature X-ray diffraction method
[p]Powdered $ZrC_{0.91}$ (contents: non-combined C – 1.0%, N – 0.34%, O – 0.23%), measured by high-temperature X-ray diffraction method
[q]Hot-pressed materials, measured by telemicroscope (dilatation) method
[r]Pyrolytic $ZrC_{\sim 1.0}$
[s]Hot-pressed $ZrC_{0.97}$, measured by dilatometric method in inert atmosphere
[t]Sintered $ZrC_{0.95}$ (contents: non-combined C – 0.14%, N – 0.26%, O – 0.11%), measured by high-temperature X-ray diffraction method
[u]Sintered (in vacuum) $ZrC_{0.96}$
[v]Hot-pressed $ZrC_{0.85}$ (contents: non-combined C – 1.4%, N – 0.94%, O – 0.18%), measured by dilatometric method
[w]Hot-pressed $ZrC_{0.95}$ (11% porosity, content non-combined C – 0.26%), measured by dilatometric method
[x]Hot-pressed and annealed (in vacuum) $ZrC_{0.98}$, measured using a high-sensitivity quartz dilatometer in inert (Ar) atmosphere
[y]Measured by dilatometric method
[z]Plasma sprayed materials, measured by dilatometric method
[a1]Sintered (in vacuum) $ZrC_{0.92}$
[a2]Powdered $ZrC_{\sim 1.0}$ (content non-combined C – 1.3%), measured by high-temperature X-ray diffraction method
[a3]Extrapolated to $ZrC_{1.0}$ composition
[a4]Hot-pressed $ZrC_{\sim 1.0}$ (content O < 0.01%), measured by high-temperature neutron diffraction method
[a5]Hot-pressed $ZrC_{0.94}$ (19–24% porosity), measured by dilatometric method
[a6]Measured by high-temperature X-ray diffraction method in high vacuum
[a7]Same for all the compositions: $ZrC_{0.72}$, $ZrC_{0.80}$, $ZrC_{0.91}$ and $ZrC_{\sim 1.0}$ (content O – 0.77%, 0.76%, 0.04%, 0.36%, respectively), measured by high-temperature X-ray diffraction method
[a8]Carburized materials, measured by telemicroscope method in inert (He) atmosphere
[a9]Pyrolytic $ZrC_{0.8–1.0}$
[b1]Measured by dilatometric method in inert atmosphere
[b2]Sintered $ZrC_{0.94}$ (content N – 0.5%), measured by dilatometric method
[b3]Hot-pressed $ZrC_{\sim 1.0}$ (3% porosity), measured by telemicroscope (dilatation) method
[b4]Powdered (annealed in vacuum) $ZrC_{\sim 1.0}$ (contents: N – 0.01%, O – 0.05%), measured by high-temperature neutron diffraction method
[b5]Hot-pressed and annealed $ZrC_{\sim 1.0}$ (3% porosity), measured by telemicroscope (dilatation) method
[b6]Grains of monocarbide phase in the arc-cast carbide-carbon eutectics in the presence of graphite flakes, measured by high-temperature X-ray diffraction method using (111) reflections
[b7]Hot-pressed $ZrC_{\sim 1.0}$ in the presence of non-combined carbon

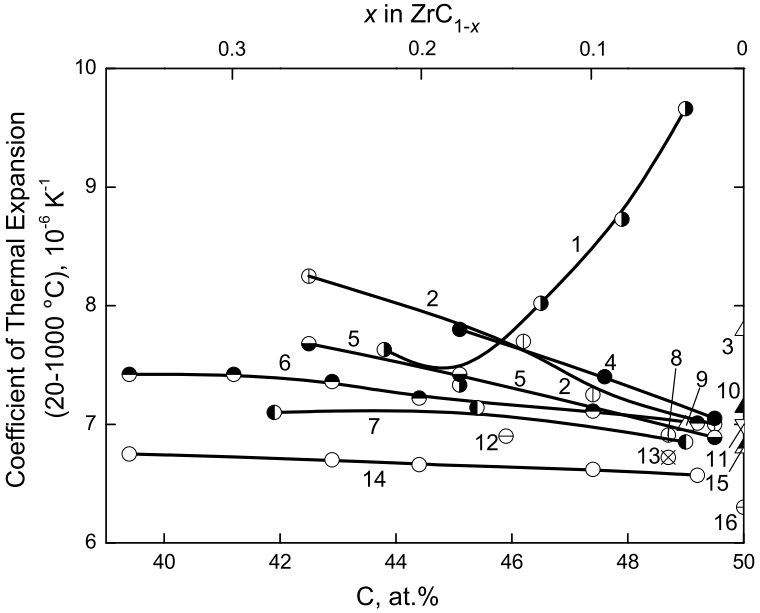

Fig. 5.10 Coefficients of linear thermal expansion in the temperature range of 20–1000 °C (when it is not indicated specially) of zirconium monocarbide ZrC_{1-x} materials (within the homogeneity range) as a function of carbide composition on the basis of several sources (1 – sintered in vacuum [4, 91, 132]; 2 – in the range of 30–330 °C, sintered in vacuum, contents: non-combined C – 0.15–0.20%, O – 0.10–0.18%, dilatometric method [277]; 3 – plasma sprayed, dilatometric method [372]; 4 – hot-pressed and annealed in vacuum, dilatometric method [384]; 5 – in the range of 20–900 °C, powdered, high-temperature X-ray diffraction method [1350]; 6 – hot-pressed, dilatometric method in inert atmosphere [1, 32, 509]; 7 – sintered in vacuum [130]; 8 – hot-pressed, dilatometric method [348]; 9 – powdered, high-temperature X-ray diffraction method [100]; 10 – [373]; 11 – [45]; 12 – hot-pressed, dilatometric method [356]; 13 – sintered, high-temperature X-ray diffraction method [60, 346]; 14 – hot-pressed, 6–10% porosity, dilatometric method in inert atmosphere [57, 364, 516]; 15 – calculated on the basis of approximation function [357]; 16 – pyrolytic [363])

and for 1500–3000 K (1230–2730 °C):

$$\Delta l/l_0 = -0.358 + (7.355 \times 10^{-4})T + (4.859 \times 10^{-8})T^2 - (6.327 \times 10^{-13})T^3, \tag{5.59}$$

where T is temperature, K. On the basis of high-temperature X-ray measurements in the temperature range of 25–2040 °C Houska [60, 346] proposed for the average coefficient of linear thermal expansion α_m, K^{-1}, of annealed (in vacuum) $ZrC_{0.95}$ (contents: non-combined C – 0.14%, N – 0.26%, O – 0.11%, Ti – 0.22%, B – 0.22%, Fe – 0.18%) the following equation:

$$\alpha_m = 6.26 \times 10^{-6} + (0.44 \times 10^{-9})(t - 25), \tag{5.60}$$

a similar linear equation for α_m, K^{-1}, of $ZrC_{0.95–0.96}$ in the temperature range of 25–2130 °C was also proposed by Katoh et al. [96]:

$$\alpha_m = 5.65 \times 10^{-6} + (1.30 \times 10^{-9})(t + 273), \tag{5.61}$$

where t is temperature, °C; according to Eq. (5.61) the coefficient of thermal expansion of monocarbide phase increases from 6.0×10^{-6} K^{-1} at room temperature to 7.3×10^{-6} K^{-1} at 1000 °C. The extrapolation of the relationship, obtained by Aigner et al. [59] for the range of 25–1200 °C, to hypothetical $ZrC_{1.0}$ phase gives the following expression for its coefficient of linear thermal expansion α, K^{-1}:

$$\alpha = \frac{2.1168 \times 10^{-6} + 1.6488 \times 10^{-9} T}{0.46896 + 2.1168 \times 10^{-6} T + 0.8247 \times 10^{-9} T^2}, \tag{5.62}$$

where T is temperature, K. The influence of carbon content on the thermal expansion of ZrC_{1-x} phase within the homogeneity range can be seen in Fig. 5.10.

The first principles calculated thermophysical properties of zirconium monocarbide materials under higher pressures are reported by Fu et al. [69], Zhu et al. [281], Varshney and Shriya [283] and Rathod et al. [379]. Some recents works [69, 167, 280, 281, 283, 284, 466] are devoted to the calculations of coefficients of thermal expansion of zirconium monocarbide by means of theoretical modelling.

In comparison with other ultra-high temperature materials the values of thermal conductivity and thermal expansion of zirconium carbide in the wide range of temperatures are summarized in Addendum.

5.3 Electro-magnetic and Optical Properties

At room temperature the values of specific electrical resistance (resistivity) of near-stoichiometric zirconium monocarbide ZrC_{1-x} lie within the area of 0.40–0.75 μΩ m [26, 27, 31–33, 74, 77, 96, 213, 307, 635, 1481]. The variations of specific electrical resistance with temperature for near-stoichiometric ZrC_{1-x} on the basis of several sources are shown in Fig. 5.11. In the wide temperature range from 0–300 °C up to 2700–3000 °C the resistance of monocarbide phase enlarges with increasing temperature, practically in accordance with linear relationship; that is an evidence of mainly metallic type of conduction in ZrC_{1-x} [26, 27]. For the temperature range of 20–2700 °C the following simple equation was recommended by Maltseva et al. [31, 213, 390] for the resistivity of near-stoichiometric zirconium monocarbide ZrC_{1-x}, ρ, μΩ m, prepared by sintering (corrected to the poreless state):

$$\rho = 0.50 + (9.0 \times 10^{-4})(t - 20), \tag{5.63}$$

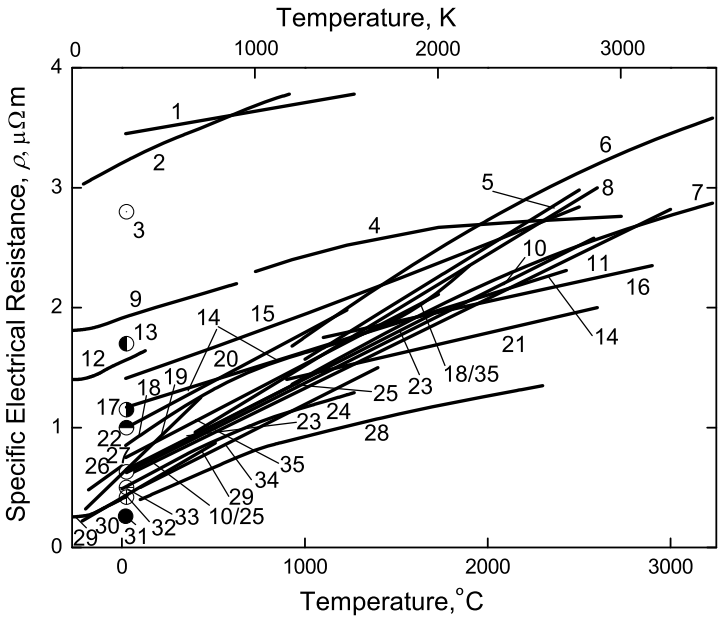

Fig. 5.11 Variation of specific electrical resistance (resistivity) with temperature for zirconium monocarbide ZrC_{1-x} materials on the basis of several sources: 1 – sintered (in vacuum) $ZrC_{0.97}$, corrected to porosity [210]; 2 – sintered and annealed (in vacuum) $ZrC_{0.97}$, corrected to porosity [405, 406]; 3 – reactively sputtered thin film, thickness 0.42 μm [414, 472]; 4 – single crystal $ZrC_{0.93}$ [396]; 5 – sintered and annealed (in vacuum) $ZrC_{\sim 1.0}$, porosity 10%, contents: non-combined C – 0.05–0.1%, N < 0.1%, O < 0.1% [117, 329]; 6 – sintered $ZrC_{\sim 1.0}$, porosity 12–13%, mean grain size – 10–20 μm, contents: non-combined C – 0.48–1.14%, N – 0.23–0.72%, O – 0.06–0.41% [397]; 7 – sintered $ZrC_{\sim 1.0}$ (*see* 6), corrected to porosity [479]; 8 – hot-pressed and annealed, content non-combined C – 1.2–1.4% [390]; 9 – single crystal $ZrC_{0.93}$ [394, 395]; 10 – hot-pressed $ZrC_{\sim 1.0}$, corrected to porosity and thermal expansion, contents: non-combined C – 0.18–0.30%, O – 0.001–0.09% [77]; 11 – recommended [33]; 12 – single crystal $ZrC_{0.98}$ [395]; 13 – sintered $ZrC_{0.98}N_{0.02}$, corrected to porosity [412]; 14 – pyrolytic $ZrC_{0.962}N_{0.035}O_{0.071}$ [413]; 15 – sintered $ZrC_{0.94}$, porosity 4%, contents: N – 0.2–0.4%, O – 0.2–0.4% [61]; 16 – hot-pressed $ZrC_{\sim 1.0}$, corrected to porosity [310, 358]; 17 – sintered (in vacuum) $ZrC_{0.96}$, porosity 5%, content N – 0.15% [399]; 18 – pyrolytic $ZrC_{0.92}$, content N – 0.8% [300]; 19 – hot-pressed $ZrC_{0.98}$, porosity 24%, contents: non-combined C – 0.3%, O – 0.05% [325]; 20 – sintered $ZrC_{\sim 1.0}$, content non-combined C – 0.3% [137]; 21 – hot-pressed $ZrC_{0.98}$, corrected to porosity [384, 387, 388]; 22 – sintered (in vacuum) $ZrC_{0.6–0.9}$, porosity 5–15%, content N – 1.4% [299]; 23 – hot-pressed $ZrC_{0.93}$, corrected to porosity [326]; 24 – sintered (in vacuum) $ZrC_{\sim 1.0}$, content N – 0.05% [398]; 25 – hot-pressed $ZrC_{0.93}$, porosity 5%, [307]; 26 – sintered, porosity 2% [321]; 27 – [35, 37, 46, 63, 407, 416]; 28 – hot-pressed $ZrC_{\sim 1.0}$, corrected to porosity [386, 402]; 29 – hot-pressed $ZrC_{0.98}$, corrected to porosity [393]; 30 – hot-pressed $ZrC_{0.97}$, porosity 9%, content O – 0.1% [325]; 31 – sintered $ZrC_{0.93}$, porosity 6–8%, content non-combined C – 0.5% [411]; 32 – [328, 392]; 33 – hot-pressed $ZrC_{\sim 1.0}$, porosity 5–7% [30, 213, 214, 382, 385, 401, 417]; 34 – hot-pressed $ZrC_{0.99}$, porosity 6%, content non-combined C – 0.12% [31, 389]; 35 – obtained from the experimental data for ZrC_{1-x}-C materials by the extrapolation to the pure carbide composition [1482] (when it is not indicated specially, data are given for near-stoichiometric compositions)

which looks similar to the proposed by Zapadaeva et al. [470]

$$\rho = 0.543 + (5.63 \times 10^{-4})(t + 273), \tag{5.64}$$

and

$$\rho = 0.393 + (7.67 \times 10^{-4})(t + 273), \tag{5.65}$$

determined by Taylor [307] for hot-pressed $ZrC_{0.93}$ and Grossman [77] for hot-pressed $ZrC_{\sim 1.0}$ materials (contents: non-combined C – 0.18–0.30%, O – 0.001–0.09%, corrected to porosity and thermal expansion) at 25–2580 °C, and also given by Katoh et al. [96] in their special review:

$$\rho = 0.372 + (8.28 \times 10^{-4})(t + 273), \tag{5.66}$$

where t is temperature, °C. At lower temperatures the resistivity of ZrC_{1-x} is perfectly described by the empirical equation

$$\rho(T) = \rho_0 + AT + B\exp(-T_0/T), \tag{5.67}$$

where T is temperature, K, ρ_0, A, B and T_0 are the constants, which were determined by Modine et al. [420] for the single crystals of $ZrC_{0.89}$ as following: ρ_0 = 1.915 μΩ m, $A = 2.5 \times 10^{-4}$ μΩ m K^{-1}, B = 0.265 μΩ m and T_0 = 495 K (ρ(300 K) = 2.04 μΩ m) (the same equation was also employed by Hinrichs et al. [396, 469] at elevated and higher temperatures (730–2730 °C) – with the values of constants: ρ_0 = 1.765 μΩ m, $A = 5.52 \times 10^{-4}$ μΩ m K^{-1}, B = 6.50 μΩ m and T_0 = 6900 K); or also by the Bloch-Grüneisen classical expression [395]

$$\rho(T) = \rho_0 + 4\rho_1 T(T/\theta)^4 J_n(\theta/T), \tag{5.68}$$

where T is temperature, K, and $J_n(\theta/T)$ is the transport integral defined by

$$J_n(\theta/T) = 2^{n-1} \int_0^{\theta/2T} x^n / \sinh^2 x dx \tag{5.69}$$

with n = 5; for the range of temperatures from −270 to +80 °C (4–350 K), the following values of ρ_0, μΩ m, ρ_1, μΩ m K^{-1}, and θ, K parameters were obtained by Modine et al. [395] from the measurements of ZrC_{1-x} single crystals with various deviations from the stoichiometry, respectively:

for $ZrC_{0.98}$: 1.397, 6.74×10^{-4}, 539;
for $ZrC_{0.93}$: 1.808, 4.90×10^{-4}, 570;
for $ZrC_{0.89}$: 1.938, 4.25×10^{-4}, 574;

at ultra-low temperatures (T < 35–50 K) the simpler expression [394]:

$$\rho(T) = \rho_0 + AT^n, \tag{5.70}$$

where A and n are constants, can be employed with n = 4.0–4.8 (the lower values of exponent n is corresponding to higher temperatures). Savvatimskiy et al. [1508] measured electrical resistance of $ZrC_{0.95}$ in the wide range of ultra-high temperatures by pulsed heating method with microsecond duration: referred to the initial linear dimensions the resistivity at 2730 °C amounted to ~3.2 ± 0.2 μΩ m, at solidus (3180 °C) and liquidus (3580 °C) temperatures it was ~2.9 ± 0.2 μΩ m and ~2.5 ± 0.2 μΩ m, respectively; with further temperature growth the resistivity of liquid zirconium carbide increased monotonically to ~3.1 ± 0.2 μΩ m at 4730 °C. The experimental data collected from the various measurements of

Table 5.8 Average values of thermal coefficients of resistivity α_R of zirconium monocarbide ZrC_{1-x} in various temperature ranges[a]

Temperature range, °C	α_R, 10^{-3} K^{-1}	Reference
(–195)–0	9.31[b]	[400]
(–183)–0	2.28[c]	[321]
(–190)–(+20)	3.22	[387]
20–950	1.23	[387]
20–1000	~1.55[d]	[32, 391]
	~0.90[e]	[32, 391]
	0.88[f]	[404]
	~0.50[g]	[32, 391]
	0.18[h]	[415]
	~0.15[i]	[32, 391]
100–1100	0.79	[389, 402, 1344]
20–1250	1.05	[387]
20–1700	0.90	[387]
20–1730	1.20	[33, 77]
400–2000	0.60	[403]
20–2150	0.79	[387]
300–2300	0.95	[26, 27, 386, 402, 1344]
20–2350	0.75	[387]
20–2700	0.90	[213]

[a]When it is not indicated specially, value reported is for near-stoichiometric composition
[b]Materials deposited from gas phase
[c]Sintered materials, porosity 2%
[d]Hot-pressed materials, carbon content – ~50 at.%, corrected to porosity
[e]Hot-pressed materials, carbon content – 47.4 at.%, corrected to porosity
[f]Hot-pressed materials, carbon content – 48.6 at.%, corrected to porosity; thermal coefficient $\alpha = 1/\rho_\theta \times (d\rho/dT)_\theta$ corresponds to Debye temperature θ = 780 K ($(d\rho/dT)_\theta = 6.6 \times 10^{-4}$ μΩ m K^{-1})
[g]Hot-pressed materials, carbon content – 44.4 at.%, corrected to porosity
[h]Thin films (2 μm thickness) prepared by activated reactive evaporation
[i]Hot-pressed materials, carbon content – 37.5 at.%, corrected to porosity

thermal coefficient of resistivity for near-stoichiometric zirconium monocarbide are listed in Table 5.8. For the phases having different deviations from the stoichiometry the variations of resistivity with temperature are presented in Fig. 5.12. The published experimental data on the resistivity of ZrC_{1-x} within the homogeneity range (Fig. 5.13) show the effect of carbon content on it as multivalued completely. Various authors [57, 85, 299, 358, 398, 408, 419] have found a decrease in resistivity values with a decrease in carbon deficit in non-metal sublattice, so Storms and Wagner [85] fit these data to the equation ($0 < x \leq 0.5$)

$$\rho = [0.382 + (0.55 + 9.50x)^{-1}]^{-1}, \tag{5.71}$$

where ρ is specific electrical resistance (resistivity), μΩ m and x is the value of index in ZrC_{1-x} formula (corresponding to *curve* 11 in Fig. 5.13); whereas other researchers [61, 134, 137, 210, 405, 406] have revealed the opposite trend in the resistivity versus carbon content relationships. However, Taylor and Storms [325] proposed later this relationship to be a dome-shaped curve with the maximum in resistivity (~2.3 μΩ m) at around $ZrC_{0.80–0.85}$ composition. In some works [61, 316, 317] it was shown that the resistivity of high-purity ZrC_{1-x} sintered materials

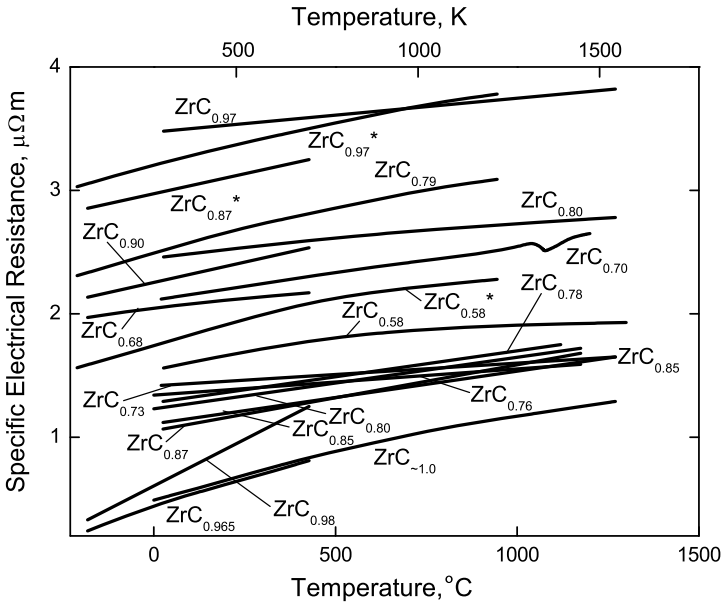

Fig. 5.12 Variations of specific electrical resistance with temperature for the ZrC_{1-x} phases having different deviations from the stoichiometry (materials sintered in vacuum, x = 0.00, 0.13, 0.15, 0.20, 0.22, 0.24, 0.27 [398]; sintered in vacuum, data corrected to porosity, x = 0.03, 0.20, 0.42 [210]; sintered in vacuum, corrected to porosity, x = 0.03*, 0.21, 0.42* [405, 406]; hot-pressed, x = 0.02 (contents: non-combined C – 0.30%, O – 0.05%), 0.035 (content O – 0.10%), 0.10 (O – 0.015%), 0.13* (O – 0.064%), 0.32 (O – 0.084%) [325]; arc-melted with order-disorder transition (superlattice formation) at 1060–1080 °C, x = 0.30 [5])

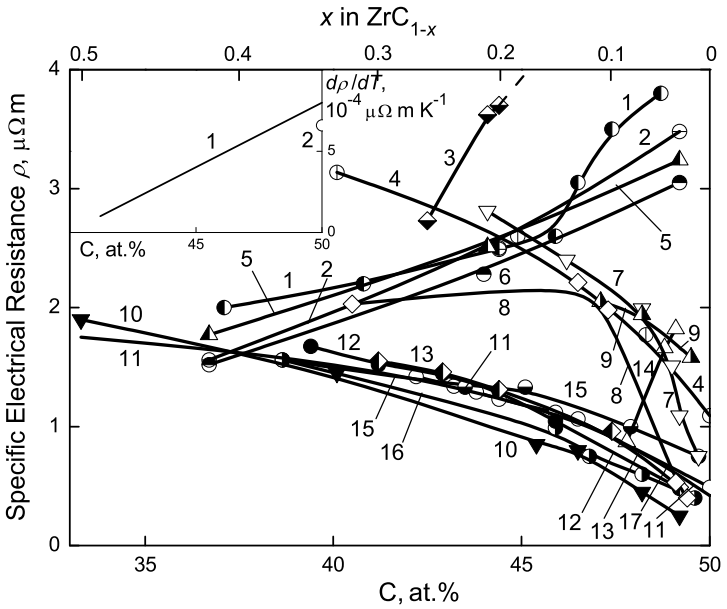

Fig. 5.13 Variations of specific electrical resistance (resistivity) at room temperature with carbon content within the homogeneity range of zirconium monocarbide ZrC_{1-x}: 1 – sintered [137]; 2 – sintered in vacuum, corrected to porosity [210]; 3 – sintered, high purity materials, 16–26% porosity [61]; 4 – hot-pressed [418]; 5 – sintered in vacuum, corrected to porosity [405, 406]; 6 – sintered in vacuum, corrected to porosity [134]; 7 – pyrolytic materials, annealed at 3000 °C [26, 27]; 8 – hot-pressed, content O – 0.015–0.10% [325]; 9 – single crystals [395]; 10 – [358]; 11 – on the basis of several sources [85]; 12 – hot-pressed [419]; 13 – hot-pressed, corrected to porosity [57]; 14 – pyrolytic, content N – 0.03–0.8% [300]; 15 – sintered, porosity 5–15%, content N > 1% [299]; 16 – [408]; 17 – [398] (*Inset* – variation of $d\rho/dT$ (slope of resistivity vs. temperature) at ambient temperatures with carbon content within the homogeneity range based on: 1 – [398], 2 – [391])

(with introduced corrections for porosity) decreases, but the resistivity of materials commercially purified increases – with the growth of vacancies in non-metal sublattice. This observation could be linked with the conclusion of independence of resistivity of carbide phases on atomic defect concentrations (carbon vacancies and atomic substitutions), which was made earlier by Golikova et al. [138, 423–425] and Avgustinik et al. [137, 426, 427]. The variation of $d\rho/dT$ (slope of resistivity vs. temperature) with ZrC_{1-x} composition at ambient temperatures is shown in the inset to Fig. 5.13. The superconducting transition temperature T_c of near-stoichiometric ZrC_{1-x} is 1.1–2.3 K [1, 421, 452, 1381]. According to Nikitin et al. [155] and Neshpor et al. [422], in the temperature range from 0.8 to 4.2 K for the ZrC_{1-x} compositions with $x < 0.36$ the temperature has little effect on the electrical conductivity, but a superconductive state was observed for materials with $x > 0.42$. At room temperature the Hall and Seebeck coefficients of near-stoichiometric zirconium monocarbide are $R = -(9–19) \times 10^{-10}$ m^3 A^{-1} s^{-1} and $S = -(10–16)$ μV K^{-1}, respectively [26, 27, 30, 131, 137, 299, 401, 405, 406, 408–410, 430, 1378]; the

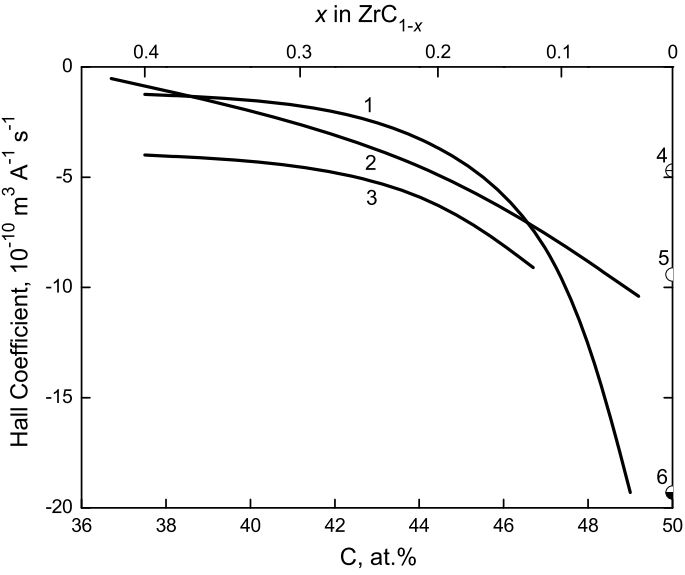

Fig. 5.14 Variations of Hall coefficient at room temperature within the homogeneity range of zirconium monocarbide ZrC_{1-x} (1 – sintered (in vacuum) materials [137]; 2 – sintered (in vacuum), corrected to porosity [1, 32, 405, 406, 409]; 3 – [131]; 4 – [432, 433]; 5 – hot-pressed [30, 401, 428, 430, 431]; 6 – [429])

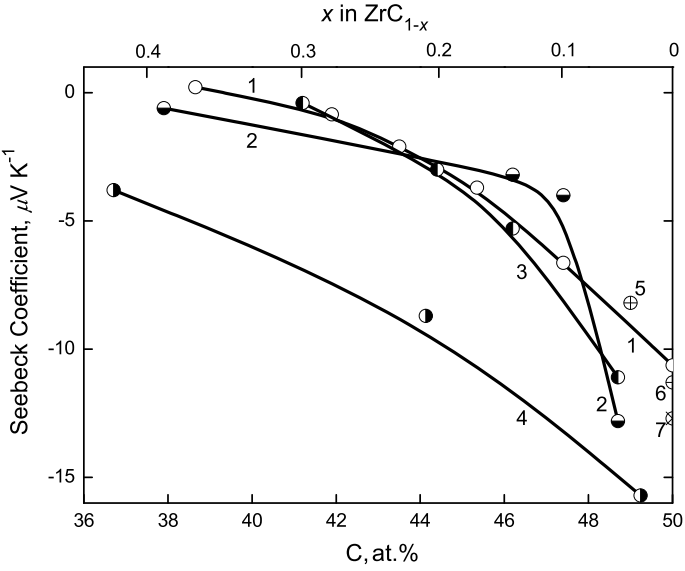

Fig. 5.15 Variations of Seebeck coefficient at room temperature within the homogeneity range of zirconium monocarbide ZrC_{1-x} (1 – sintered materials [299]; 2 – sintered (in vacuum) [137]; 3 – [408]; 4 – sintered (in vacuum) [405, 406, 409]; 5 – pyrolytic, content N – 0.03–0.08% [300]; 6 – [430]; 7 – hot-pressed [401, 428])

experimental data obtained by several researchers for these coefficients varying within the homogeneity range of ZrC_{1-x} are presented in Figs. 5.14 and 5.15.

Near-stoichiometric zirconium monocarbide ZrC_{1-x} is a diamagnetic substance with molar magnetic susceptibility χ_m (SI) $\approx -(30–300) \times 10^{-6}$ cm^3 mol^{-1} at room temperature slightly increasing with temperature in the range from 20 to 1000 °C [26, 27, 154, 437–439]. However, an increase in carbon deficit (value of x) in ZrC_{1-x} results in changing in the magnetic properties of zirconium monocarbide phase from diamagnetic to paramegnetic, so the maximum of molar magnetic susceptibility χ_m (SI) $\approx (740–770) \times 10^{-6}$ cm^3 mol^{-1} is corresponding to the lowest carbon content in the homogeneity range, and χ_m (SI) ≈ 0 – to the composition of $\sim ZrC_{0.8–0.9}$, as it is shown in Fig. 5.16.

The room-temperature reflectance spectrum of sintered near-stoichiometric ZrC_{1-x} (1% porosity, mean surface roughness $R_a = 23 \pm 2$ nm, distance between the highest peak and the lowest valley $R_t = 0.68 \pm 0.08$ μm) from the ultraviolet wavelength region to the mid-infrared band is shown in Fig. 5.17. The reflectance of single-crystalline $ZrC_{0.98}$ in the interval of 0.025–6.5 eV was measured by Modine et al. [420], and that of hot-pressed $ZrC_{\sim 1.0}$ in the interval of 1.0–8.0 eV – by Alward et al. [1486] and reactively sputtered $ZrC_{\sim 1.0}$ thin films – by Bonnot et al. [414, 472]. The soft X-ray reflectivity data of thin films deposited by ion beam sputtering were reported by Singh et al. [198, 1430]; there it was shown that the presence of non-combined carbon and oxygen contamination lower the reflectivity

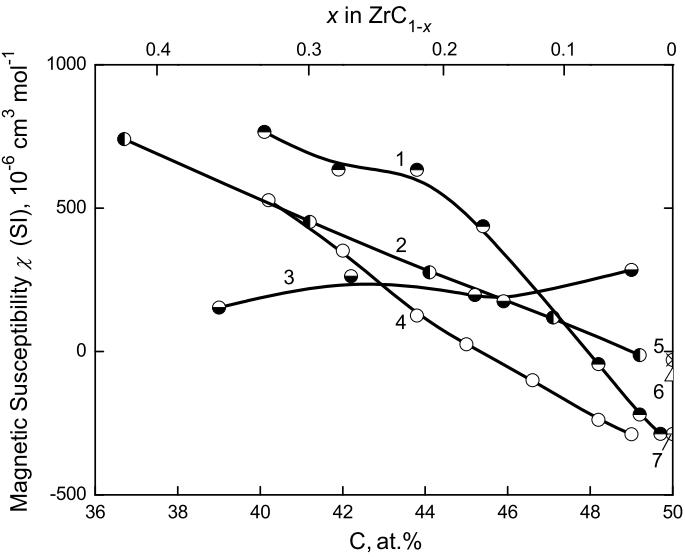

Fig. 5.16 Variation of molar magnetic susceptibility at room temperature within the homogeneity range of zirconium monocarbide ZrC_{1-x} on the basis of several sources: 1 – [434]; 2 – contents: N – 0.04%, Fe – 0.02% [154]; 3 – [136]; 4 – [159]; 5 – [409, 436–439]; 6 – extrapolated value [154]; 7 – [433, 435]

performance of ZrC_{1-x}. Delin et al. [1487] presented the results of *ab initio* calculations of optical properties for quasi-stoichiometric zirconium monocarbide. Optical spectra in the infrared (IR) and visible ranges as well as X-ray emission and absorption spectra of ZrC_{1-x} were considered by Zhurakovskii et al. [442, 443], Ramqvist et al. [444], Kosolapova et al. [163], Guerrero et al. [445] and Kammori et al. [1488] and summarized by Upadhyaya [27]. The absorption wave numbers of basic maxima of IR band for zirconium monocarbide ZrC_{1-x} are 1078 cm^{-1} (for $ZrC_{0.96}$) [163], 1055 cm^{-1} (for $ZrC_{0.70}$) [163] and 776, 1332, 1383 and 1634 (for $ZrC_{\sim 1.0}$) [445]; the latter 4 baselines are used for the FTIR quantitative determination of ZrC_{1-x} phase in ceramic materials. At the common conditions the colour of zirconium monocarbide ZrC_{1-x} materials is dark-grey in the dispersed state (powdered) or steel/silver-grey with characteristic metallic glitter in the compact state [1, 11, 30–32, 63, 452]; this is related to the broad selection of possible electron transitions in the entire spectral range of visible light and the small area of the free Fermi surface corresponding to non-localized electrons. Such "absence" of colour is inherent for ZrC_{1-x} in the entire homogeneity region, although near its lower boundary the materials take on a lighter (more silvery) colour; it occurs due to the increasing overlap of the valency band with the conduction band and the increase in free Fermi surface, which causes some selectivity of reflections [26, 27].

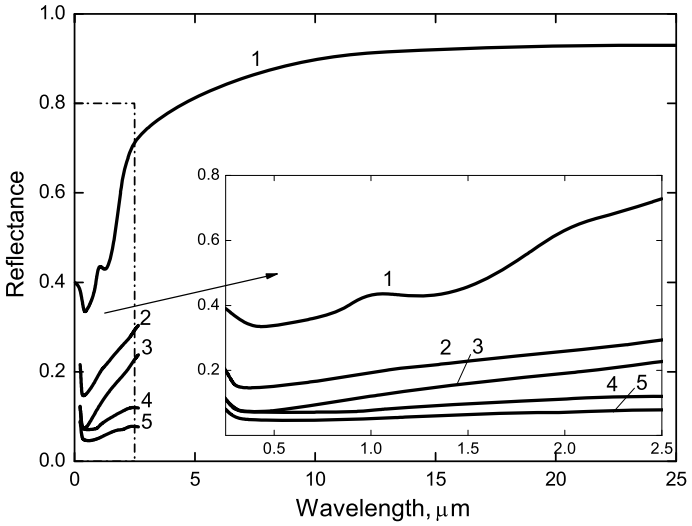

Fig. 5.17 Room-temperature normal reflectance spectrum of near-stoichiometric ZrC_{1-x} from the ultraviolet wavelength region to the mid-infrared band: 1 – sintered materials (1% porosity, mean surface roughness $R_a = 23 \pm 2$ nm, distance between the highest peak and the lowest valley $R_t = 0.68 \pm 0.08$ μm) [440, 441]; 2 – powder, cold-pressed [468]; 3 – powder, cold-pressed and heated at 700 °C, 0.5 h [468]; 4 – powder, cold-pressed and heated at 900 °C, 1 h [468]; 5 – powder, cold-pressed and heated at 900 °C, 0.5 h [468]

Table 5.9 Normal monochromatic emittance ε_λ (λ = 0.650–0.665 μm) of near-stoichiometric zirconium monocarbide ZrC_{1-x} materials in various temperature ranges

Monochromatic emittance, ε_λ	Temperature range, °C	Reference
0.43–0.56[a]	1050–2100	[467]
0.49	2300–2500	[275]
0.50–0.70[b]	1600–2500	[26, 454]
0.51[c]	2600–3300	[450]
0.51–0.54[d]	2200–3100	[465]
0.51–0.83	1500–3300	[459]
(0.53 ± 0.05)–(0.55 ± 0.05)[e]	2200–3200	[453]
(0.55 ± 0.05)–(0.68 ± 0.06)[f]	900–2200	[453]
0.55–0.72[g]	1700–2600	[455]
0.58[h]	900–2700	[1, 446]
0.58–0.90[i]	900–3100	[479]
0.59–0.61[j]	2000–2600	[470]
0.60–0.65	800–2100	[460]
0.62–0.65[k]	730–2230	[33]
0.63 ± 0.02[l]	800–2100	[77]
0.63–0.88	1100–3200	[449]
0.64–0.80	1400	[461]
0.69–0.91[m]	850–3100	[26]
0.69–0.94[n]	900–3200	[397, 447]
0.71[o]	2000–2900	[26]
0.75–0.79[p]	800–2000	[30–32, 451]
0.75–0.85[q]	1300–2800	[465]
0.76–0.86[r]	1500–2100	[275]
0.78[s]	1500	[478]
0.79–0.80[t]	1100–1700	[26]
0.80–0.83[c]	1500	[450]
0.81–0.85[d]	1300–1900	[465]
0.90[u]	1400–1900	[473]
0.97	700–1800	[448]

[a]Arc-melted (via floating zone) polycrystalline $ZrC_{0.92\pm0.02}$, measured in vacuum 10^{-5}–10^{-6} Pa; ε_λ falls with the temperature growth linearly (slope is ~1.2×10^{-4} K^{-1})

[b]Sintered materials; ε_λ falls with the temperature growth linearly

[c]Hot-pressed $ZrC_{0.98}$ (porosity $\leq$ 10%); with a hysteresis effect of ~400 °C, depending to some extent of the heating/cooling rate, observed from 1500 to 2600 °C (probably due to the surface oxidation and/or nitridation)

[d]Sintered $ZrC_{0.95}$ (contents before measurement: non-combined C < 0.1%, O < 0.05%), measured in high-purity He; ε_λ falls with the temperature growth linearly

[e]Analyzed on the basis of several sources: measurements from polished and stabilized surfaces, ε_λ falls very slightly with the temperature growth (almost linearly)

[f]Analyzed on the basis of several sources: measurements from polished and stabilized surfaces, ε_λ falls with the temperature growth (almost linearly)

[g]ε_λ falls with the temperature growth

[h]Pyrolytic $ZrC_{0.86–0.99}$ coating (contents: non-combined C – traces, N $\leq$ 1%, O – 0.2–0.4%)

[i]Sintered $ZrC_{\geq 0.99}$ (13% porosity, mean grain size – 10–20 μm, contents: non-combined C – 0.48–1.14%, N – 0.23–0.72%, O – 0.06–0.41%) measured in vacuum and Ar multiply; ε_λ falls with the temperature growth

[j]Pyrolytic $ZrC_{0.99}$ (contents: non-combined C – 0.03%, N – 0.15%, O – 0.28%)

[k]Recommended for high-temperature technological applications

(continued)

Table 5.9 (continued)

[l]Hot-pressed (in vacuum) $ZrC_{\sim 1.0}$ (contents: non-combined C – 0.18–0.30%, O – 0.001–0.09%)
[m]Sintered materials; ε_λ falls with the temperature growth linearly
[n]Sintered $ZrC_{\sim 1.0}$ (12–13% porosity, contents: non-combined C – 0.3–1.1%, N – 0.5–0.7%, O – 0.4%); ε_λ falls with the temperature growth linearly
[o]Hot-pressed materials
[p]Powdered materials (coatings); ε_λ increases slightly with the temperature growth
[q]Sintered $ZrC_{0.95}$ (contents before measurement: non-combined C < 0.1%, O < 0.05%), measured in commercial He; ε_λ falls with the temperature growth linearly
[r]ε_λ falls with the temperature growth
[s]Powdered materials (coatings) on tantalum substrate
[t]Hot-pressed materials; ε_λ falls with the temperature growth
[u]For $\lambda = 0.81$ μm, the value is constant and rather independent on surface roughness and thermal treatment in vacuum and hot hydrogen

The normal monochromatic emittance (spectral emissivity) ε_λ of zirconium monocarbide ZrC_{1-x} materials slightly varies with temperature, the linear relationship for $\lambda = 0.65$ μm

$$\varepsilon_\lambda = 0.715 - (1.174 \times 10^{-4})(t + 273), \tag{5.72}$$

was obtained by Mackie et al. [467, 469] for arc-melted (via floating zone) polycrystalline monocarbide $ZrC_{0.92\pm0.02}$ by vacuum (10^{-5}–10^{-6} Pa) measurements in the temperature range 1050–2100 °C. Similar equations were proposed by Naumenko [513] for 730–2200 °C

$$\varepsilon_\lambda = 0.80 - (2.86 \times 10^{-5})(t - 727) \tag{5.73}$$

and by Rithof et al. [454] for 1600–2500 °C

$$\varepsilon_\lambda = 0.90 - (2.29 \times 10^{-4})(t - 727), \tag{5.74}$$

where t is temperature, °C. At the ZrC_{1-x}-C eutectic temperature (~2880 °C) the normal monochromatic emittance ε_λ ($\lambda = 0.65$ μm) of zirconium monocarbide $ZrC_{\sim 1.0}$ phase is close to 0.6 [99]. For the integral (total) hemispherical emittance ε_T of hot-pressed (in vacuum) $ZrC_{\sim 1.0}$ (8.5% porosity), Grossman [77] also proposed the linear equation (for 1200–2100 °C)

$$\varepsilon_T = 0.772 - (9.0 \times 10^{-5})(t + 273), \tag{5.75}$$

where t is temperature, °C. According to the experimental works by Naumenko [513], emissivity ε_λ of ZrC_{1-x} phases decreases from 0.78 to 0.72 with increasing value of index x in ZrC_{1-x} formula from 0.03 to 0.30. The data on normal monochromatic emittance ε_λ ($\lambda = 0.650$–0.665 μm) and integral (normal and hemispherical) emittances ε_T for ZrC_{1-x} materials produced by different manufacturing methods and measured in various temperature intervals are listed in

Table 5.10 Integral (total) emittance ε_T of near-stoichiometric zirconium monocarbide ZrC_{1-x} in various temperature ranges

Integral emittance, ε_T	Temperature range, °C	Reference
0.25–0.72[a]	800–1150	[476]
0.35–0.51[b]	900–2700	[1, 446]
0.37–0.40[c]	950–1400	[26, 1489]
0.38–0.55[d]	900–2600	[463, 464]
0.38–0.56[e]	1000–3000	[479]
0.38–0.83[f]	900–2700	[456, 471]
0.40–0.47[g]	2000–2600	[470, 477]
0.41–0.57[h]	900–3100	[449]
0.42–0.46[i]	1650–2200	[457]
0.43–0.44[j]	2000–2500	[454]
0.43–0.45[k]	1800–2600	[458]
0.43–0.46[l]	2000–2500	[455]
0.50[l]	2000–3100	[275]
0.50–0.68[m]	730–2730	[33]
0.55–0.57[n]	2000–2900	[26]
0.56–0.64[o]	1200–2100	[77]
0.58–0.62[p]	1250–2100	[460]
0.59[q]	3530	[477]
0.64–0.85[r]	900–3100	[397]
0.70–0.87[s]	600–1400	[480, 481]
0.89–0.91[t]	1100	[474, 475]

[a]Hemispherical emittance ε_T for hot-pressed $ZrC_{\sim 1.0}$ (porosity < 2%, mean grain size – 10–20 μm, contains non-combined C); ε_T increases with the temperature growth (almost linearly)

[b]Hemispherical emittance ε_T for pyrolytic $ZrC_{0.86–0.99}$ coating (contents: non-combined C – traces, N $\leq$ 1%, O – 0.2–0.4%); ε_T increases with the temperature growth

[c]Hemispherical emittance ε_T for sintered materials; ε_T falls with the temperature growth

[d]Hemispherical emittance ε_T for pyrolytic materials (content O after the measurement dropped from 0.06 to 0.04%); ε_T increases with the temperature growth linearly

[e]Hemispherical emittance ε_T for Sintered $ZrC_{\geq 0.99}$ (13% porosity, mean grain size – 10–20 μm, contents: non-combined C – 0.48–1.14%, N – 0.23–0.72%, O – 0.06–0.41%) measured in vacuum and Ar multiply; curve $\varepsilon_T - T$ with the minimum at ~1400 °C and maximum at ~2500 °C

[f]Normal emittance ε_T for sintered materials, measured in Ar; ε_T increases slightly with the temperature growth up to ~1900 °C and then falls sharply with further temperature increase

[g]Normal emittance ε_T for pyrolytic $ZrC_{0.99}$ (contents: non-combined C – 0.03%, N – 0.15%, O – 0.28%); ε_T increases with the temperature growth (almost linearly)

[h]Hemispherical emittance ε_T for sintered materials; curve $\varepsilon_T - T$ with the minimum at ~1400 °C and maximum at ~2500 °C

[i]Normal emittance ε_T for hot-pressed and polished materials, measured in Ar, integrated from spectral data; ε_T increases with the temperature growth

[j]Normal emittance ε_T for sintered materials; ε_T increases with the temperature growth linearly

[k]Normal emittance ε_T for $ZrC_{\sim 1.0}$ with content non-combined C $\geq$ 2.5%, integrated from spectral data; ε_T increases with the temperature growth

[l]Hemispherical emittance ε_T

[m]Recommended for high-temperature technological applications; ε_T falls with the temperature growth

[n]Normal emittance ε_T for hot-pressed; ε_T increases with the temperature growth

[o]Hemispherical emittance ε_T for hot-pressed (in vacuum) $ZrC_{\sim 1.0}$ (8.5% porosity, contents: non-combined C – 0.18–0.30%, O – 0.001–0.09%); ε_T falls with the temperature growth linearly

(continued)

Table 5.10 (continued)

[p]Hemispherical emittance ε_T for hot-pressed $ZrC_{\sim 1.0}$ (10% porosity, contents: non-combined C $\geq$ 0.16%, O < 0.002%), measured in vacuum ~10^{-4} Pa; ε_T falls with the temperature growth

[q]Theoretically calculated normal emittance ε_T for stoichiometric composition

[r]Hemispherical emittance ε_T for sintered $ZrC_{\sim 1.0}$ (12–13% porosity, contents: non-combined C – 0.3–1.1%, N – 0.5–0.7%, O – 0.4%); ε_T falls with the temperature growth from 0.82 up to the minimum at ~2700 °C (almost linearly) and then increases to the maximum with the further temperature growth

[s]Hemispherical emittance ε_T for sintered pigmented coating, measured in vacuum < 7×10^{-4} Pa; ε_T increases with the temperature growth

[t]Hemispherical emittance ε_T for 25–50 μm thickness vacuum-plasma sprayed coating (after annealing ε_T falls to 0.72–0.86 depending on substrate compositions)

Tables 5.9 and 5.10. The variations of normal monochromatic emittance $\varepsilon_{\lambda,T}$ with wavelength and temperature for ZrC_{1-x} are shown in Fig. 5.18. Thermodynamics of thermal radiation for stoichiometric zirconium monocarbide was considered in details by Fisenko et al. [477, 1490]; the same authors studied thermal radiation of solid and liquid zirconium monocarbide [1428] on the basis of experimental data for the normal spectral emissivity in a visible-near infrared range (0.55 μm $\leq \lambda$ 0.90 μm) at melting (eutectic) point obtained by Manara et al. [99].

The thermionic emission characteristics (electron work function and Richardson constants) of various zirconium monocarbide ZrC_{1-x} phases are given in Table 5.11 and Fig. 5.19. The calculated values of emission current density for near-stoichiometric monocarbide ZrC_{1-x} are about 56, 1.5×10^3 and 1.9×10^5 A m^{-2} at 730, 1230 and 1730 °C, respectively [33]. The work functions found for ZrC_{1-x} thin films in cesium vapour are 0.9–1.1 eV higher than those given in literature due to

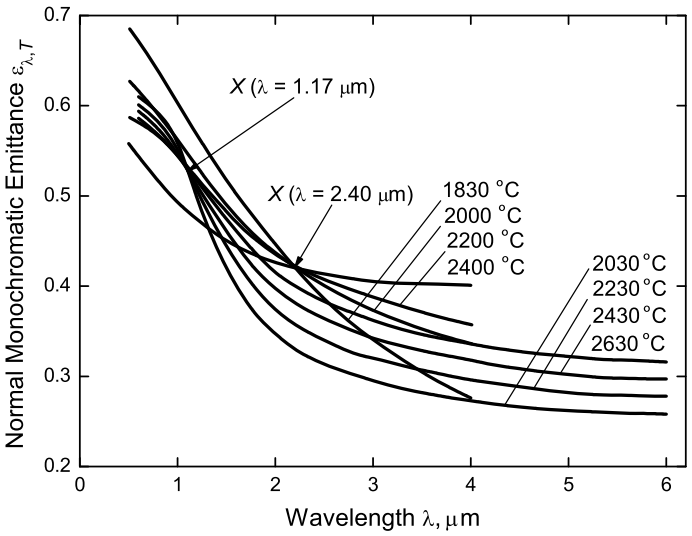

Fig. 5.18 Variations of normal monochromatic emittance $\varepsilon_{\lambda,T}$ with wavelength λ for zirconium monocarbide materials: sintered $ZrC_{\sim 1.0}$ (96% purity, contains non-combined C) – at 1830, 2000, 2200 and 2400 °C [454, 462]; pyrolytic $ZrC_{0.99}$ (contents: non-combined C – 0.03%, N – 0.15%, O – 0.28%) – at 2030, 2230, 2430 and 2630 °C [470] (for the families of ε – λ curves *X*-points or isosbestic points, where $|1/\varepsilon_{\lambda,T} \times d\varepsilon_{\lambda,T}/dT|_{\lambda=X} = 0$, are marked specially)

Table 5.11 Thermionic emission characteristics (electron work function and Richardson constant) of zirconium monocarbide ZrC_{1-x} phases

Composition	Work function[a], $\varphi = \varphi_0 + (d\varphi/dT)_{av}T$, eV	Richardson constant, A, 10^4 A m^{-2} K^{-2}	Temperature range, °C	Remarks[b]	Reference
$ZrC_{0.70}$	3.88[c]	–	1500	TiM, hot-pressed	[26, 27, 513]
	$5.10 - (6.8 \times 10^{-4})T$	–	1350–1800		[1, 485]
$ZrC_{0.74}$	4.0[d]	–	–	PhES, nanocrystalline thin films	[1491]
$ZrC_{0.75}$	3.95[c]	–	1500	TiM, hot-pressed	[26, 27, 513]
	$5.10 - (6.6 \times 10^{-4})T$	–	1300–1800	TiM, hot-pressed	[1, 485]
$ZrC_{0.80}$	3.97[c]	–	1500	TiM, hot-pressed	[26, 27, 513]
	$5.17 - (6.7 \times 10^{-4})T$	–	1300–1800	TiM, hot-pressed	[1, 485]
$ZrC_{0.84}$	3.40 ($d\varphi/dT = 0$)	–	1000–1700	CPDM, single crystal (102), non-combined C – 0.8%, measured in vacuum (~1.3×10^{-6} Pa)	[482, 483, 521, 522, 563]
	3.49[c]	–	1200–1500	CPDM, single crystal (100), non-combined C – 0.8%	[521, 522]
	3.50 ($d\varphi/dT = 0$)	–	1000–1700	TiM, single crystal (102), non-combined C – 0.8%, measured in vacuum (~1.3×10^{-6} Pa)	[563]
	$3.54 - (2.45 \times 10^{-5})T$	–	≤1700	TiM, single crystal (100), non-combined C – 0.8%	[482, 483, 521, 522]
$ZrC_{0.88}$	4.39[c]	–	1500	TiM, hot-pressed	[1, 141]
$ZrC_{0.90}$	4.00[c]	–	1500	TiM, hot-pressed	[26, 27, 513]
	$5.10 - (6.0 \times 10^{-4})T$	–	1300–1800	TiM, hot-pressed	[1, 485]
$ZrC_{0.91\pm0.02}$	3.25	–	–	Single crystal (210), theoretically calculated	[526]
	3.26	–	–	Single crystal (110), theoretically calculated	[526]
	3.26–3.49	–	1600–2100	TiM, single crystal, (210) plane within 6°	[523, 524]
	3.32[c]	–	1300	TiM (projection microscope), single crystal (210)	[526]
	3.33–3.34[c]	–	1700	TiM, single crystal (210)	[523, 525, 526]
	3.35	–	1400–1700	TiM, single crystal (100)	[523, 524]
	3.46[c]	–	1300	TiM (projection microscope), single crystal (110)	[526]

(continued)

Table 5.11 (continued)

Composition	Work function[a], $\varphi = \varphi_0 + (d\varphi/dT)_{av}T$, eV	Richardson constant, A, 10^4 A m^{-2} K^{-2}	Temperature range, °C	Remarks[b]	Reference
	3.48–3.62	–	1600–1900	TiM, single crystal (100), cleaned surface	[523, 524]
	3.50[c]	–	–	TiM, single crystal, average value	[526]
	3.53	–	–	Single crystal (100), theoretically calculated	[526]
	3.56–3.61	–	1800–2100	TiM, single crystal (100)	[523, 524]
	3.59	–	1800	TiM, single crystal, (311) plane, ~7° off axis	[523, 524]
	3.59[c]	–	1500	FEM, single crystal (100)	[527]
	3.60[c]	–	1700	TiM, single crystal (100)	[523, 525]
	3.61–3.77	–	1700–2250	TiM, single crystal (100)	[523, 524]
	3.63[c]	–	1300	TiM (projection microscope), single crystal (100)	[526]
	3.80	–	–	Single crystal (111), theoretically calculated	[526]
	3.85[c]	–	1300	TiM (projection microscope), single crystal (111)	[526]
$ZrC_{0.94}$	3.72[c]	–	1400	TiM, contents: non-combined C – 1.15%, N – 1.48%, O $>$ 0.4%	[1, 482, 497, 498]
$ZrC_{0.96-0.97}$	$4.82 - (4.3 \times 10^{-4})T$	–	–	TiM, pyrolytic, textured surface with (100) preferential orientation	[482, 484]
$ZrC_{0.97}$	3.95 ± 0.04	–	1700	TiM, hot-pressed	[484, 485]
	4.07[c]	–	1500	TiM, hot-pressed	[26, 27, 513]
	4.28 ± 0.05[e]	–	25	CPDM, hot-pressed	[484, 485]
	4.43 ± 0.04	–	25	TiM, hot-pressed	[484, 485]
	$4.52 - (2.8 \times 10^{-4})T$	–	1300–1750	TiM, hot-pressed	[1, 484, 485]
$ZrC_{\sim 1.0}$	1.60	–	–	TiM, in the presence of Cs vapour	[26, 27, 495, 500]
	2.10[c]	–	–	TiM, powder	[1, 486]
	2.10[c]	–	900–1200	TiM, on W substrate, in the presence of Cs vapour	[504, 529]
	2.16	–	–	TiM, powder on W substrate	[507, 508, 529]
	2.18	–	–	TiM	[1, 482]
	2.18[c]	0.31	900–1700	TiM, powder on W substrate	[1, 487, 529]
	2.30	0.20	–	TiM, powder on Ta substrate	[1, 488]

(continued)

Table 5.11 (continued)

Composition	Work function[a], $\varphi = \varphi_0 + (d\varphi/dT)_{av}T$, eV	Richardson constant, A, 10^4 A m^{-2} K^{-2}	Temperature range, °C	Remarks[b]	Reference
	2.60[c]	1.12	1400–1700	TiM, in the presence of Cs vapour	[1, 33, 489, 529]
	2.70	0.15	–	TiM, powder on Ta substrate	[1, 482, 490]
	2.86	88.15	1200–2200	TiM	[1, 482]
	2.95[c]	–	25	FEM, thin films on Mo substrate, measured in vacuum (~3×10^{-8} Pa)	[529, 531, 1493]
	2.96	14	–	TiM, powder on Ta substrate	[1, 33, 482]
	3.00	0.7	–	TiM, powder on Ta substrate with $MoSi_2$ underlayer	[1, 482, 490]
	3.00	–	–	TiM, pyrolytic coating on Ta wire	[1, 259, 482]
	3.10	36	–	TiM, hot-pressed	[1, 482]
	3.1 ± 0.1	0.5 ± 0.1	–	TiM, melted in Ar	[1, 482]
	3.15[c]	–	25	On Si substrate, measured in vacuum (~3×10^{-8} Pa)	[529, 531]
	3.2 ± 0.1	3 ± 2	–	TiM, sintered, 8% porosity	[1, 482]
	3.2 ± 0.1	8 ± 2	–	TiM, sintered, 15% porosity	[1, 482]
	3.20[c]	–	800	Measured in vacuum (~5×10^{-5} Pa of O_2)	[529]
	$3.24 + (2.0 \times 10^{-4})T$	–	1000–1700	TiM, powder	[1, 482]
	$3.24 + (2.0 \times 10^{-4})T$	–	1000–2500	TiM, pyrolytic coating on W substrate	[1, 33, 482, 491]
	3.30[c]	–	1700	TiM	[1, 482]
	3.32	17	1300–1800	TiM	[1, 482]
	3.4 ± 0.1	16 ± 6	–	TiM, sintered, 44% porosity	[1, 482]
	3.49[c]	–	1500	TiM, powder on Ta substrate	[1, 33, 482]
	3.50	–	–	AeE	[1, 482, 1422, 1426]
	3.5[c]	–	–	FEM, single crystal tips (field emitters)	[1493]
	3.54[c]	–	25	FEM, thin films on W substrate, measured in vacuum (~3×10^{-8} Pa)	[529, 531, 532, 1493]
	3.56[c]	181	2200	TiM, powder on Ta substrate	[535]
	3.58[c]	–	2100	TiM, hot-pressed	[1, 492]
	3.58[c]	–	25	On Mo substrate, measured in vacuum (~3×10^{-8} Pa)	[529, 532]

(continued)

Table 5.11 (continued)

Composition	Work function[a], $\varphi = \varphi_0 + (d\varphi/dT)_{av}T$, eV	Richardson constant, A, 10^4 A m^{-2} K^{-2}	Temperature range, °C	Remarks[b]	Reference
	3.60[c]	–	1900	TiM, powder on W substrate	[1, 482, 493]
	3.6[c]	–	–	TiM, sintered, average value accepted	[526]
	3.62 ± 0.01[c]	–	25	CPDM (Mo), electrophoretic coating on Ta substrate, measured in vacuum (~10^{-7}–10^{-8} Pa)	[1, 33, 482, 494, 529]
	3.64[c]	–	1700	TiM, powder on W substrate	[1, 482]
	3.66	–	1700	TiM, sintered	[523, 524]
	3.67[c]	–	1000–1300	Measured in vacuum ($\leq 4\times10^{-7}$ Pa)	[517, 519, 529]
	3.70[c]	–	2050	TiM, powder on Ta substrate	[1, 482]
	3.70[c]	–	2050	TiM, melted	[1, 482]
	3.70	32	–	TiM, pressed	[1, 482]
	3.70	–	–	TiM, cleaned surface in the presence of Cs vapour	[1, 482, 495]
	3.70 ± 0.04[c]	–	1100–1200	TiM, high purity materials, measured in vacuum (~10^{-7} Pa)	[1, 495, 496, 529]
	3.70–3.95		700–1100	TiM, high purity materials, measured in vacuum (~10^{-7} Pa)	[496, 529]
	3.73[f]	–	–	Single crystal (001), theoretically evaluated	[710]
	3.74 ± 0.07	–	–	Recommended as the most probable value for essentially clean surfaces	[529]
	3.75[c]	–	1000–1700	TiM, sintered, 15% porosity	[1, 482, 529]
	~3.75	–	1700	TiM, hot-pressed, 5–9% porosity	[1379]
	3.78[c]	–	1000–1700	TiM, sintered, 8–45% porosity	[1, 26, 27, 482, 529]
	3.80	134	1300–1700	TiM, melted	[1, 33, 499, 529]
	3.80	–	–	TiM, sintered rod	[1, 500]
	3.80	–	830–1130	TiM	[1, 501]
	3.80	–	1700	TiM, carbide-graphite hypereutectic alloy	[1, 482, 502]
	$3.82-(6.35\times10^{-4})T$	–	700–1300	Measured in vacuum ($\leq 10^{-7}$ Pa)	[529, 533]
	3.85[c]	–	–	TiM, single crystal (100) film on W (100) substrate	[528]
	3.90[c]	–	2100	TiM, powder on Ta substrate	[1, 482, 500]

(continued)

Table 5.11 (continued)

Composition	Work function[a], $\varphi = \varphi_0 + (d\varphi/dT)_{av}T$, eV	Richardson constant, A, 10^4 A m^{-2} K^{-2}	Temperature range, °C	Remarks[b]	Reference
	3.90[c]	–	2200	TiM, powder on W substrate	[1, 482, 503]
	3.94	–	–	Calculated value (unrelaxed surface)	[1422, 1426]
	3.95 ± 0.04[c]	–	1300	TiM, hot-pressed, measured in vacuum (<10^{-6} Pa)	[482, 504, 529]
	3.97[c]	–	1700	TiM	[516]
	3.97[g]	–	–	Single crystal (001), theoretically evaluated	[710]
	4.00[c]	140	–	TiM, melted, measured in vacuum (~10^{-5} Pa)	[1, 33, 505, 529]
	4.00[c]	–	1500	TiM, on Mo/Ta substrates	[1, 482, 529]
	4.0[h]	–	25	FEM, single crystal (001), clean surface	[557–560, 1492]
	4.01 ± 0.05[c]	–	25	CPDM, hot-pressed, plain, measured in vacuum (<3×10^{-7} Pa)	[482, 504, 529]
	4.02[c]	–	1650	TiM, melted in Ar	[1, 26, 27, 482]
	4.09[c,i]	–	–	Single crystal (100), theoretically evaluated	[529, 530]
	4.19[c]	–	25	FEM, thin films on Si substrate, measured in vacuum (~3×10^{-8} Pa)	[529, 532, 1493]
	4.20	360	1100–1800	TiM, electrophoretic coating on W filament	[1, 482, 506]
	4.30	–	–	Calculated value (relaxed surface)	[1422, 1426]
	4.30[c]	–	800	Measured in vacuum (~10^{-3} Pa of O_2)	[529]
	4.39[c]	–	1500	–	[26, 27, 141]
	4.40–4.80	–	–	TiM, single crystal (111) film on W (110) substrate	[528]
	4.45[c,j]	–	–	Single crystal (100), theoretically evaluated	[529, 530, 1422, 1426]
	4.48[c]	–	1300–1600	–	[529, 534]
	4.52	–	–	FEM, thin films on W substrate, clean surface	[1493]
	4.60	–	–	FEM, thin films on Mo substrate, clean surface	[1493]
	4.7[k]	–	25	FEM, single crystal (111), clean surface	[557–560]

(continued)

Table 5.11 (continued)

Composition	Work function[a], $\varphi = \varphi_0 + (d\varphi/dT)_{av}T$, eV	Richardson constant, A, 10^4 A m^{-2} K^{-2}	Temperature range, °C	Remarks[b]	Reference
	4.82	–	–	FEM, thin films on Si substrate, clean surface	[1493]
	–	–	1700	TiM, powder on Ta, current density: 10^4 A m^{-2} (constant mode) and 5×10^4 A m^{-2} (impulse mode)	[1, 482, 488]
	–	–	–	FEM, current density: (1.2–6.2) $\times$ 10^4 A m^{-2} (impulse mode)	[527]

[a]T is temperature, K
[b]Methods applied for the experimental determination of the work function (TiM – thermionic method, AeE – autoelectron emission, FEM – field emission microscopy, PhES – photoelectron spectroscopy, CPDM – contact potential difference method, second electrode is given in brackets) and manufacturing methods for the fabrication of a particular material (or its constitution) are marked
[c]Values of effective electron work function
[d]The presence of amorphous C and/or metallic Zr is very probable
[e]For materials with additions of 0.5–1.0% Y φ = 5.3–5.6 eV
[f]First principles calculated using the Perdew's generalized gradient approximation (GGA) for the fixed surface model
[g]First principles calculated using the Perdew's GGA for the relaxed surface model
[h]For graphene-covered surface φ = 4.2 eV
[i]Calculated using the full-potential linear-muffin-tin-orbital (FP-LMTO) method
[j]Calculated using the linear-muffin-tin-orbital with atomic sphere approximation (LMTO-ASA) technique
[k]For graphene-covered surface φ = 4.3 eV

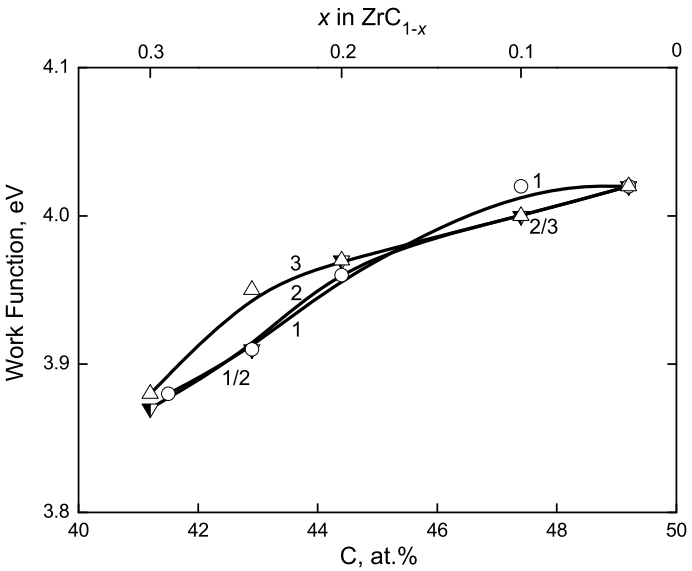

Fig. 5.19 Variation of effective work function of hot-pressed zirconium monocarbide ZrC_{1-x} materials with carbon content measured by "total current" method at 1530 °C: 1 – [482, 485, 516], 2 – [57, 510–512, 514, 515, 1502], 3 – [26, 27, 513]

uncontrolled impurity gases containing oxygen [517–519]. Using the photoemission technique, Lindberg and Johansson [1387] revealed that the work function of $ZrC_{0.92}$ (100) increases dramatically with increasing the exposure to O_2, while only small increases in the work function are observed after the exposure to CO media. Similar to some other transition metal carbides, single crystal ZrC_{1-x} materials show the following relationships between effective thermionic work functions of different crystallographic planes: $\varphi_{(210,\ 310)} < \varphi_{(100)} \approx \varphi_{(110)} \gg \varphi_{(111)}$ [527]. Field emission properties of ZrC_{1-x} single crystal tips and thin films (coatings) were reported in several works [527, 531, 532, 558, 1491, 1493–1501]. Secondary ion emission of hot-pressed ZrC_{1-x} materials was studied by Cherepin et al. [520].

The recommended values of electrical resistivity, magnetic susceptibility, integral and spectral emittances and thermionic emission characteristics (electron work function and Richardson constant) for zirconium monocarbide are given in the wide range of temperatures in comparison with other ultra-high temperature materials in Addendum.

5.4 Physico-mechanical Properties

The physico-mechanical properties of zirconium monocarbide ZrC_{1-x} are sensitive to the deviations from the stoichiometry, but also to the crystallographic directions in the materials. At room temperature the hardness *HV*, GPa of near-stoichiometric zirconium monocarbide ZrC_{1-x} is evaluated as (17–20) ± (2–3) [581], 17.9–28.7 [63], (17.9–18.0) ± (0.6–1.0) (9.8 N load, mean grain size – 5 μm) [441, 587, 588], 19.6–27.5 [540], 20–34 [74], 20.5 (porosity – 3–5%, 0.5 N load) [538, 556], 21.0 (for $ZrC_{0.98}$, porosity 9%, mean grain size – 50 μm) [537], (21.6–22.6) ± (1.7–2.2) [1, 26, 27, 547, 548], 21.7 (contents: non-combined C – 0.05–1.0%, O ≤ 0.01%, N ≤ 0.01%, porosity – 10%, mean grain size – 10 μm) [117], 22.7–24.7 (0.5 N load) [553], 22.8 (contents: non-combined C – 0.05–1.0%, O ≤ 0.1%, N ≤ 0.1%, porosity – 10%, mean grain size – 10 μm) [117], 23.5 (1.2 N load) [213], 24.0 (for $ZrC_{0.96}$, 1 N load) [1, 4, 132], 24.5 [1, 539], 24.5 (for $ZrC_{0.79}$) [104], 25.0 ± 1.5 (for $ZrC_{0.96}$, contents: O – 0.01%, N – 0.46%) [4, 545], 25.4 (0.5 N load) [213], 25.5 (0.5–2.0 N load) [43, 46, 214, 218, 384, 550, 552], 25.5 (for $ZrC_{0.98}$, 200 N load) [28, 546], 25.7 (for $ZrC_{0.96}$) [554], 25.9 [452], 26.5 [557], 26.5–28.4 [47], 27.0 (porosity – 7%, mean grain size – 10–14 μm) [541], 27.5 (2 N load) [48], 27.5–30.9 [103], 27.8 ± 0.9 [35, 542, 549], 27.8–34.1 (1 N load) [39, 550], 28.0 [351, 536], 28.6 ± 1.8 (0.3 N load) [30, 33, 213], 28.7 [30, 47, 213], 28.9 ± 1.2 (for $ZrC_{0.97}$, 0.5 N load) [32, 513], 29.3 [555] and 32.9 (for $ZrC_{0.92}$) [300]; hardness *HK*, GPa is 17.9 (1 N load) [319, 543, 544], 21.0 [39, 551] and 25.5 (0.5 N load) [319, 543]. The hardness of near-stoichiometric zirconium monocarbide ZrC_{1-x} in Mohs scale is >8 [63], 8–9 [32, 43, 547, 1511], or 9 [46], and in Rockwell scale *HRA* it is 87 kgf mm^{-2} (0.85 GPa) [4, 30, 33, 48, 547], 88–94 kgf mm^{-2} (0.86–0.92 GPa) [46], or 92.5 kgf mm^{-2} (0.91 GPa) [39]. According to the data available in literature within the homogeneity region the microhardness $H\mu$ of ZrC_{1-x} enlarges almost linearly from low-carbon to high-carbon phase boundaries in the range from 18.7 to 26.5 GPa (0.5 N load) [79], or from 19.6 to 28.4 GPa [1506]. The variations of hardnesses *HV*/*HK* and microhardness $H\mu$ with carbon content within the homogeneity range of ZrC_{1-x},

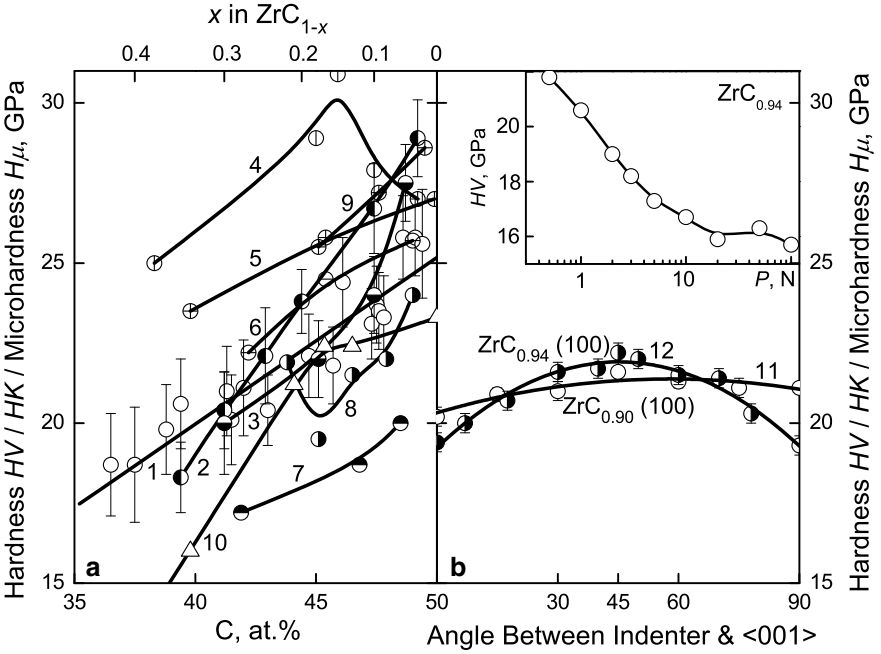

Fig. 5.20 Variations of the hardnesses *HV* (1, 10–11), *HK* (4, 12) and microhardness *Hμ* (2–3, 5–9, 11) at room temperature for prepared by various methods zirconium monocarbide ZrC_{1-x} materials: (**a**) with deviation from the stoichiometry within the homogeneity range (1 – 200 N load [28, 546]; 2 – sintered (in vacuum) materials, contents: O – 0.08–0.17%, N – 0.01–0.09%, 0.5 N load [32, 513, 516]; 3 – fused (in Ar) materials, 1 N load [561]; 4 – thin films prepared by tri-ion beam-assisted deposition [103]; 5 – carbidized coatings on graphite produced by soaking graphite in liquid metal, 0.5 N load [615]; 6 – fused (in Ar) materials [554]; 7 – hot-pressed, 0.2 N load [562, 590]; 8 – sintered (in vacuum) materials, porosity – 6–30%, mean grain size – 10–18 μm, contents: non-combined C – 0.10–0.55%, O – 0.13–0.72%, N – 0.01–0.30%, 1 N load [132]; 9 – hot-pressed (annealed in vacuum) materials, 0.5 N load [384]; 10 – DFT-estimated [696]) and (**b**) with indenter orientation (angle between indenter axis and <001> direction) on the (100) surface of single crystal materials grown by floating-zone processes (11 – $ZrC_{0.90}$, 2 N load [93]; 12 – $ZrC_{0.94}$, 5 N load [171, 1503, 1504]); *Inset* – variation of hardness with test force loading for spark plasma sintered $ZrC_{0.94}$, porosity – 1%, contents: non-combined C – 0.28%, O – 0.79%, N – 0.52%, Fe – 0.02% in *HV*, GPa – lg*P*, N scale [726]

indenter orientation (for the (100) surface of $ZrC_{0.90}$ and $ZrC_{0.94}$ single crystals) and test force loading are demonstrated in Fig. 5.20. In the wide temperature range from 0 to 1400 °C the hardness of ZrC_{1-x} materials can be expressed by the following exponential temperature dependency [591–593]:

$$HV = 29.4 \exp[-(16.4 \times 10^{-4})t], \tag{5.76}$$

where *HV* is Vickers hardness, GPa and *t* is temperature, °C, which is approximated at lower and moderate temperatures (<600 °C) by the linear equation [54]:

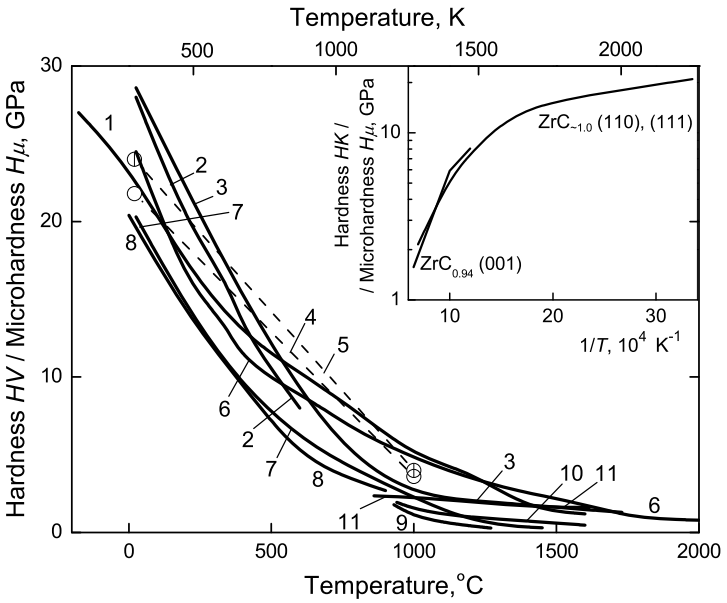

Fig. 5.21 Variations of the hardness HV (3, 6, 8–10) and microhardness $H\mu$ (1–2, 4–5, 7) of zirconium monocarbide ZrC_{1-x} materials with temperature: 1 – $ZrC_{0.95}$, zone melted single crystal (100), contents: non-combined C – 0.05%, O – 0.01%, N – 0.03%, W – 0.46%, 1.15 N load [537]; 2 – $ZrC_{\sim 1.0}$, sintered [536]; 3 – recommended for near-stoichiometric compositions by Kotelnikov et al. [33]; 4 – $ZrC_{0.78}$, sintered (in vacuum) and annealed, porosity – 21%, mean grain size – 18 μm, contents: non-combined C – 0.1%, O – 0.26%, N – 0.20%, 1 N load [132]; 5 – $ZrC_{0.96}$, sintered (in vacuum) and annealed, porosity – 6%, mean grain size – 10 μm, contents: non-combined C – 0.55%, O – 0.13%, N – 0.07%, 1 N load [132]; 6 – $ZrC_{\sim 1.0}$, hot-pressed and annealed (in H_2), porosity – 1–2%, content non-combined C – 0.54%, 9.8 N load [573–575, 595]; 7 – $ZrC_{0.90}$, grown by radio-frequency floating-zone process (zone melted) single crystal (100), 2 N load [172]; 8 – $ZrC_{\sim 1.0}$, [567]; 9 – $ZrC_{0.75}$, fused (in Ar) [568]; 10 – $ZrC_{0.82}$, fused (in Ar) [568]; 11 – $ZrC_{\sim 1.0}$, sintered, porosity – 7%, 1.5 kN load [1412] (*Inset* – single crystal materials: $ZrC_{0.94}$ (001), grown by the floating-zone technique, 5 N load in lg(HK) – 1/T, K scale [572] and $ZrC_{\sim 1.0}$ (110), (111) in lg($H\mu$) – 1/T, K scale [333])

$$HV = 34.9 - (3.53 \times 10^{-3})t. \tag{5.77}$$

The variations of the hardness HV and microhardness $H\mu$ of near-stoichiometric zirconium monocarbide with temperature on the basis of several sources are presented in Fig. 5.21. The data on microhardness $H\mu$ of zirconium monocarbide single crystals and various polycrystalline materials are listed in Tables 5.12 and 5.13.

On the basis of several sources the variations of ultimate tensile strength σ_t (some data on σ_t are supplied with corresponding elongation δ and reduction in area ψ values) with temperature of near-stoichiometric zirconium monocarbide ZrC_{1-x}, which were determined on materials manufactured by various techniques, are shown in Fig. 5.22 (not given in the figure, the highest value of tensile strength of ZrC_{1-x} materials $\sim$0.7 GPa (at room temperature) [1] was revealed for carbide polycrystalline fibre prepared by the impregnation of viscose rayon with zirconium chloride solution followed by the heat treatment). For ZrC_{1-x} materials the stress-strain diagrams (at moderate strain rates $\leq 10^{-3}$ s^{-1}) under tension and compression are linear up to 1200 and 2000 °C, respectively; at higher strain rate ($\geq 10^{-1}$ s^{-1}) ZrC_{1-x}

Table 5.12 Microhardness $H\mu$ of zirconium monocarbide ZrC_{1-x} ($0 < x \leq 0.16$) single crystals at room temperature

Surface	Indentor diagonal direction	Microhardness $H\mu$, GPa	Reference
		$x \approx 0$	
(100)[a]	<100>	26.0	[1, 564]
	<110>	25.0	[1, 564]
	<210>	27.0	[1, 564]
(110)[a]	<100>	25.0	[1, 564]
	<110>	24.0	[1, 564]
	<111>	26.0	[1, 564]
	<112>	25.5	[1, 564]
(001)	<100>	19.4	[1470, 1505]
	<110>	22.1	[1470, 1505]
		$x = 0.04$[b]	
(100)	<110>	25.0	[1, 564]
	<210>	27.0	[1, 564]
(110)	<100>	26.0	[1, 564]
	<110>	24.0	[1, 564]
	<111>	26.0	[1, 564]
	<112>	25.5	[1, 564]
(111)	<100>	25.0	[1, 564]
	<110>	23.0	[1, 564]
	<112>	24.0	[1, 564]
	<123>	25.0	[1, 564]
		$x = 0.05$[c]	
(100)	–	24.0	[1, 537]
		$x = 0.06$	
(100)	<100>	19.8[d]	[171]
		19.4 ± 0.3[e]	[171]
	<110>	20.9[d]	[171]
		22.2 ± 0.3[e]	[171]
	<010>	19.3 ± 0.3[e]	[171]
	<$\bar{1}$00>	19.6 ± 0.3[e]	[171]
	<$\bar{1}$10>	22.1 ± 0.3[e]	[171]
	–	21.2[d]	[572]
		$x = 0.08$–0.11[f]	
(100)	–	24.0–26.0	[571]
		$x = 0.10$	
(100)	<100>	17.2 ± 0.8[g]	[570]
	<001>	20.3[h]	[4, 93]
	<010>	17.2 ± 0.8[g]	[570]
		20.6[h]	[4, 93]
	<110>	18.7 ± 0.8[g]	[570]
	<011>	21.6[h]	[4, 93]
	–	26.0	[566]

(continued)

Table 5.12 (continued)

Surface	Indentor diagonal direction	Microhardness $H\mu$, GPa	Reference
(110)	<001>	18.8[h]	[172]
(111)	<110>	17.4[h]	[172]
		x = 0.11	
(001)	–	25.0 ± 0.5[i]	[565]
	–	24.6 ± 0.5[j]	[565, 577]
(110)	–	22.9 ± 0.5[j]	[565, 577]
	–	22.8 ± 0.5[i]	[565]
(111)	–	23.8 ± 0.5[i]	[565]
	–	23.5 ± 0.5[j]	[565, 577]
		x = 0.16[k]	
(100)	–	24.2 ± 2.9	[1, 563]

[a]Crystals produced by induction zone melting, 1 N load, Knoop measurement (*HK*)
[b]Crystals produced by induction zone melting, contents: non-combined C – 0.05%, O – 0.01%, N – 0.01%, 1 N load, Knoop measurement (*HK*)
[c]Crystals grown by controlled solidification from the melt, contents: non-combined C – 0.05%, O – 0.01%, N – 0.03%, W – 0.46%, 1.15 N load
[d]Crystals grown by the floating zone technique, 5 N load
[e]Crystals grown by the floating zone technique, 5 N load, Knoop measurement (*HK*)
[f]Crystals produced by induction zone melting, contents: non-combined C – 0.05–0.59%, O – 0.01–0.95%, N – 0.02–0.12%, W – 0.45–0.52%, 1 N load
[g]Crystals produced by zone melting, 2 N load
[h]High-purity crystals grown by radio-frequency floating-zone process, 2 N load
[i]Crystals grown by controlled solidification from the melt, 1 N load, Knoop measurement (*HK*)
[j]Crystals grown by controlled solidification from the melt, annealed at 2050 °C, 1 N load, Knoop measurement (*HK*)
[k]Crystals produced by plasma-arc melting, content: non-combined C – 0.8%

Table 5.13 Microhardness $H\mu$ of the various compositions of zirconium monocarbide ZrC_{1-x} phase at room temperature

Composition	Microhardness $H\mu$, GPa	Remarks	Reference
$ZrC_{0.54}$	18.2	0.5 N load, sintered	[35, 79]
$ZrC_{0.65}$	23.5	0.5 N load, carbidized coating on graphite produced by soaking graphite in liquid metal	[615]
	18.3 ± 1.1	0.5 N load, sintered in vacuum, contents: O – 0.14%, N – 0.08%	[32, 513]
$ZrC_{0.66}$	16.0	DFT-estimated	[696]
$ZrC_{0.70}$	20.4 ± 1.2	0.5 N load, sintered in vacuum, contents: O – 0.13%, N – 0.06%	[32, 513]
	20.0 ± 1.6	1 N load, fused in Ar	[1, 561]
$ZrC_{0.72}$	17.2	0.2 N load, hot-pressed	[1, 562]

(continued)

Table 5.13 (continued)

Composition	Microhardness $H\mu$, GPa	Remarks	Reference
$ZrC_{0.75}$	22.1 ± 1.5	0.5 N load, sintered in vacuum, contents: O – 0.12%, N – 0.09%	[32, 513]
$ZrC_{0.77}$	27.8 ± 0.9	0.2–1.5 N load, hot-pressed and annealed, porosity – 12%, contents: non-combined C – 0.2%	[35, 542, 549]
$ZrC_{0.78}$	21.9[a]	1 N load, sintered in vacuum and annealed, porosity – 21%, mean grain size – 18 μm, contents: non-combined C – 0.1%, O – 0.26%, N – 0.20%	[1, 4, 132]
$ZrC_{0.79}$	24.5[b]	Reactive magnetron sputtered thin films, thickness <3 μm, mean grain size – 10–13 nm, hardness values were taken at ~0.1–0.2 μm depth	[104]
	21.2	DFT-estimated	[696]
$ZrC_{0.79}O_{0.02}$	27.0	0.3 kN load, hot-pressed, porosity <0.5%, mean grain size – ~0.1 μm	[175, 179]
$ZrC_{0.79}O_{0.13}$	21.0 ± 1.4[b]	Spark plasma sintered, porosity < 1%, mean grain size – 5.2 ± 0.2 μm	[630]
$ZrC_{0.80}$	23.8 ± 1.0	0.5 N load, sintered in vacuum, contents: O – 0.17%, N – 0.07%	[32, 513]
$ZrC_{0.82}$	22.0 ± 1.2	1 N load, fused in Ar	[1, 561]
	19.5[c]	1 N load, sintered in vacuum and annealed, porosity – 30%, mean grain size – 12 μm, contents: non-combined C – 0.47%, O – 0.72%, N – 0.30%	[1, 4, 132]
$ZrC_{0.83}$	25.8	0.5 N load, carbidized coating on graphite produced by soaking graphite in liquid metal	[615]
	22.4	DFT-estimated	[696]
$ZrC_{0.87}$	22.4	DFT-estimated	[696]
	21.5	1 N load, sintered in vacuum and annealed, porosity – 15%, mean grain size – 15 μm, contents: non-combined C – 0.10%, O – 0.62%, N – 0.01%	[1, 132]
$ZrC_{0.88}$	18.7	0.2 N load, hot-pressed	[1, 562]
$ZrC_{0.89}$	24.1	0.5 N load, sintered	[79]
$ZrC_{0.90}$	26.7 ± 1.4	0.5 N load, sintered in vacuum, contents: O – 0.11%, N – 0.03%	[32, 513]
	24.0 ± 1.2	1 N load, fused in Ar	[1, 561]
	(17.8–18.5) ± (0.4–0.5)	9.8 N load, reactive hot-pressed, porosity – ~1%, mean grain size – 50–100 μm	[1425]
$ZrC_{0.92}$	32.9	Pyrolytic materials, in the parallel direction to the chemical vapour deposition surface	[4, 300]
	22.0	1 N load, sintered in vacuum and annealed, porosity – 14%, mean grain size – 11 μm, contents: non-combined C – 0.10%, O – 0.50%, N – 0.01%	[1, 4, 132]
$ZrC_{0.93\pm0.01}$	16.0[d]	2 N load, hot-pressed, porosity – 7%, mean grain size – 4 μm, contents: non-combined C – 0.85%, O – 0.56%, N – 0.32%, Hf – 1.0%	[356, 634]
$ZrC_{0.94}$	20.0	0.2 N load, hot-pressed	[1, 562]
$ZrC_{0.94}O_{0.05}$	28.3 ± 1.0[b]	Spark plasma sintered, porosity <1%, mean grain size – 4.7 ± 0.2 μm	[630]

(continued)

Table 5.13 (continued)

Composition	Microhardness $H\mu$, GPa	Remarks	Reference
$ZrC_{0.95}$	27.5 ± 1.2	1 N load, fused in Ar	[1, 561]
	24.7 ± 1.0	0.5 N load, hot-pressed and work-hardened by grinding, porosity – 2–3%, contents: non-combined C – 0.91%, Fe – 0.30%	[553]
	22.0	0.5 N load, extruded, presintered and sintered non-isothermally, porosity – 14%, mean grain size – 11 μm, contents: non-combined C – 0.04%, O – 0.33%, N – 0.03%, W – 0.4%	[623]
	20.5 ± 1.0	0.5 N load, hot-pressed and recrystallized at 1680 °C, porosity – 2–3%, contents: non-combined C – 0.91%, Fe – 0.30%	[553]
	18.0	0.5 N load, extruded, presintered and sintered non-isothermally, porosity – 4–8%, mean grain size – 16–30 μm, contents: non-combined C – 0.04%, O – 0.33%, N – 0.03%, W – 0.4%	[623]
$ZrC_{0.96}$	25.7	Fused materials	[554]
	25.0 ± 1.5	1 N load, produced by carbidization through metal saturation, contents: O – 0.01%, N – 0.46%	[1, 4, 545]
	24.0[e]	1 N load, sintered in vacuum and annealed, porosity – 6%, mean grain size – 10 μm, contents: non-combined C – 0.55%, O – 0.13%, N – 0.07%	[1, 4, 132]
	(14.4–16.2) ± (0.3–0.9)	9.8 N load, reactive hot-pressed, porosity < 2%, mean grain size – 5–10 μm, trace amount of non-combined C	[1425]
$ZrC_{0.97}$	28.9 ± 1.2	0.5 N load, sintered in vacuum, contents: O – 0.08%, N – 0.01%	[32, 513]
$ZrC_{0.98}$	27.2[b]	Hot-pressed, mean grain size – 5–20 μm	[703]
	26.5 ± 0.6	Pyrolytic materials	[579]
	23.2	Hot-pressed, mean grain size – 5–20 μm	[703]
	21.0	Sintered materials, porosity – 9%, mean grain size – 50 μm	[537]
$ZrC_{\sim 1.0}$	48 ± 3[b]	5–750 μN load, pulsed laser deposited thin films, mean grain size – 19 nm, microstrain – 1.6%, content O < 0.5%	[729]
	41 ± 2[b]	5–750 μN load, pulsed laser deposited thin films, mean grain size – 12 nm, microstrain – 1.7%, content O < 0.5%	[729]
	29.3	Hot-pressed materials	[555]
	28.7	–	[30, 213]
	28.6 ± 1.8	0.3 N load	[30, 33]
	28.4	Value extrapolated to the stoichiometric composition (high-carbon ZrC_{1-x} phase boundary)	[35, 1506]
	28.0	–	[351, 536]
	27.9–28.4	–	[635]
	27.8–34.1	1 N load	[39, 550]
	27.8	0.5 N load, sintered	[35, 542, 1509]

(continued)

Table 5.13 (continued)

Composition	Microhardness $H\mu$, GPa	Remarks	Reference
	27.5–30.9	Thin films prepared by tri-ion beam-assisted deposition	[103]
	27.5	2 N load	[48]
	27.0	0.5 N load, carbidized coating on graphite produced by soaking graphite in liquid metal	[336, 615]
	26.5–28.4	–	[47]
	26.5	0.5 N load, sintered, porosity – 7%, mean grain size – 10–14 μm	[1, 541, 557]
	26.5	Value extrapolated to the stoichiometric composition (high-carbon ZrC_{1-x} phase boundary)	[35, 79]
	26.2–30.5	Calculated on the basis of hardness-elasticity correlation	[578, 586]
	26.0[f]	0.2–10 N load, hot-pressed and annealed, porosity <5%	[547, 569, 589, 591]
	26.0	Value summarized on the basis of critical review of available experimental data	[65]
	26.0 (24.0[d])	Electron beam deposited thin films on Ti substrate, thickness – 0.51 ± 0.02 μm, mean grain size <0.2 μm	[580]
	25.8	–	[719]
	25.5	0.5–2.0 N load	[43, 46, 54, 260, 550, 552]
	25.5[d]	0.5 N load	[319, 543, 1412]
	25.4	0.5 N load	[213]
	25.2 ± 1.4[b]	Spark plasma sintered, porosity – 1–2%, mean grain size – 5–13 μm	[588]
	25.0–25.9	On the basis of several sources	[582]
	24.5	Pyrolytic materials	[1, 539]
	23.7	4.9 N load, spark plasma sintered, porosity <3%, mean grain size – ~30 μm	[1431]
	23.5	1.2 N load	[213]
	23.3	DFT-estimated	[696]
	23.0	Pyrolytic materials	[363]
	22.9	4.9 N load, spark plasma sintered, porosity <4%, mean grain size – 15 μm	[1431]
	22.8	Sintered materials, porosity – 10%, mean grain size – 10 μm, contents: non-combined C – 0.05–1.0%, O ≤ 0.1%, N ≤ 0.1%	[117]
	22.3[g]	High-pressure hot-pressed, porosity – 6%	[576]
	22.2	Hot-pressed, porosity – 1–2%, content non-combined C – 0.54%	[573, 574]
	21.8	0.5 N load, spark plasma sintered, porosity – 1%	[726]
	21.7	Sintered materials, porosity – 10%, mean grain size – 10 μm, contents: non-combined C – 0.05–1.0%, O ≤ 0.01%, N ≤ 0.01%	[117]

(continued)

Table 5.13 (continued)

Composition	Microhardness $H\mu$, GPa	Remarks	Reference
	21.6[h]	–	[1510]
	21.3	4.9 N load, spark plasma sintered, porosity <5%, mean grain size – ~10 μm	[1431]
	21.0[d]	–	[39, 551]
	21.0	Calculated theoretically on the basis of Tian's model	[582, 584, 585]
	20.9	–	[721]
	20.6	1 N load, spark plasma sintered, porosity – 1%	[726]
	20.5	0.5 N load, sintered, porosity – 3–5%	[1, 538]
	20 ± 1[b]	5–750 μN load, pulsed laser deposited thin films, mean grain size – 6 nm, microstrain – 2.0%, content O < 0.5%	[729]
	19.6–27.5	Pyrolytic materials, carbide is in contact with carbon	[540]
	19.3	Calculated on the basis of Gao's model	[582, 583]
	19.0	2 N load, spark plasma sintered, porosity – 1%	[726]
	18.6[i]	1.5 load, sintered and annealed (in Ar), grain size from 10 to 60 μm	[616]
	18.0 ± 0.4	9.8 N load, hot-pressed, porosity – 2%	[1479]
	(17.9–18.0) ± (0.6–1.0)	9.8 N load, spark plasma sintered, porosity – 1–2%, mean grain size – 5–13 μm	[441, 587, 588]
	17.9–28.7	General estimation	[63]
	17.9[d]	1 N load	[319, 543, 544]
	17.3	4.9 N load, spark plasma sintered, porosity – 1%	[726]
	(17–20) ± (2–3)	Pulsed laser ablation deposited thin films on Ti substrate, thickness – 0.59–0.60 μm, approximate grain size – 10–100 nm	[581]
	16.7	9.8 N load, spark plasma sintered, porosity – 1%	[726]
	16.3 ± 1.6	0.1 kN load, reactive spark plasma sintered materials, porosity – 4%, mean grain size <10 μm	[594]
	15.7	0.1 kN load, spark plasma sintered, porosity – 1%	[726]
	15.6	DFT-estimated	[376]

[a]3.6 GPa – at 1000 °C
[b]Evaluated by nanoindentation with a diamond tip
[c]3.9 GPa – at 1000 °C
[d]Knoop measurement (*HK*)
[e]4.0 GPa – at 1000 °C
[f]Determined in accordance with the formula: $H\mu = 18.18P/d^2$, where P is the load on the indentor, N and d is the diagonal of the indentation, μm; the values of other micromechanical characteristics: microbrittleness $\gamma_\mu = (D^2 - d^2)/d^2 = 6.0$, where D is the average size of the damageability zone, μm and brittle microstrength $\sigma_\mu = P/D^2 = 2.1$ GPa [569]
[g]Depth-sensing indentation measurement
[h]Diamond pyramid hardness measurement (*DPH*)
[i]Independent on grain size and annealing temperature

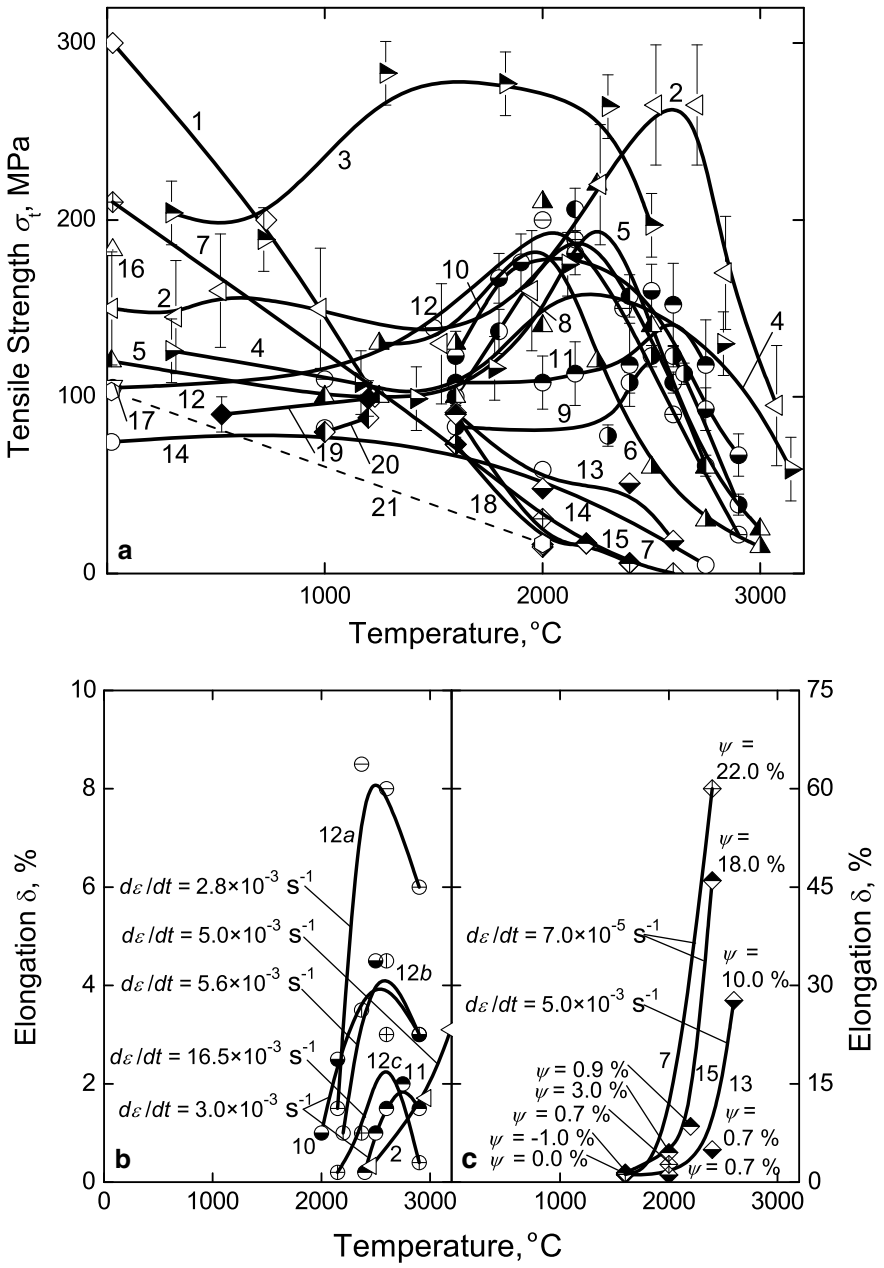

Fig. 5.22 (see caption on page 486)

◂**Fig. 5.22** Variation of (**a**) ultimate tensile strength σ_t, (**b**, **c**) elongation δ and (**c**) reduction in area ψ at fracture with temperature for zirconium monocarbide ZrC_{1-x} materials on the basis of several sources: 1 – sintered $ZrC_{\sim 1.0}$, porosity – 9% [30, 33, 607]; 2 – sintered $ZrC_{\sim 1.0}$, porosity – 5%, strain rate $d\varepsilon/dt = 5 \times 10^{-3}$ s^{-1} [333, 336]; 3 – sintered and treated thermomechanically $ZrC_{0.97}$, porosity – 5% [632, 633]; 4 – same, without special thermomechanical treatment [632, 633]; 5 – sintered $ZrC_{0.95}$, porosity – 7%, mean grain size – 8 μm, contents: O + N – 0.14%, $d\varepsilon/dt = 3 \times 10^{-3}$ s^{-1} [4, 645]; 6 – same, $d\varepsilon/dt = 3 \times 10^{-5}$ s^{-1} [4, 645]; 7 – hot-pressed $ZrC_{0.83}$, porosity – 9%, mean grain size – 6 μm, contents: non-combined C – 1.40%, O – 0.18%, N – 0.94%, $d\varepsilon/dt = 7 \times 10^{-5}$ s^{-1} [356, 634]; 8 – cold-rolled and sintered $ZrC_{0.94}$, porosity – 14%, mean grain size – 12 μm, contents: non-combined C – 0.10%, O – 0.08%, N – 0.02%, W – 0.20, $d\varepsilon/dt = 3 \times 10^{-3}$ s^{-1} [597]; 9 – cold-rolled and sintered $ZrC_{0.95}$ (zonal segregation structured), porosity – 6%, mean grain size – 10 μm, contents: non-combined C – 0.10%, O – 0.20%, W – 0.20, $d\varepsilon/dt = 3 \times 10^{-3}$ s^{-1} [597]; 10 – cold-pressed and sintered $ZrC_{0.95}$, porosity – 7%, mean grain size – 8 μm, contents: non-combined C – 0.10%, O – 0.12%, N – 0.02%, W – 0.40, $d\varepsilon/dt = 3 \times 10^{-3}$ s^{-1} [597]; 11 – cold-pressed and sintered $ZrC_{0.98}$ (zonal segregation structured), porosity – 6%, mean grain size – 12 μm, contents: non-combined C – 0.60%, O – 0.08%, N – 0.008%, W – 0.30, $d\varepsilon/dt = 3 \times 10^{-3}$ s^{-1} [597]; 12 – sintered $ZrC_{\sim 1.0}$, porosity – 5%, mean grain size – 15–25 μm, $d\varepsilon/dt = 2.8 \times 10^{-3}$ s^{-1} [601, 620, 645]; 13 – hot-pressed and annealed $ZrC_{0.83}$, porosity – 9%, mean grain size – 6 μm, contents: non-combined C – 1.40%, O – 0.18%, N – 0.94%, $d\varepsilon/dt = 5 \times 10^{-3}$ s^{-1} [356, 634]; 14 – sintered $ZrC_{\sim 1.0}$, porosity – 30% [32, 596]; 15 – hot-pressed $ZrC_{0.93\pm0.01}$, porosity – 7%, mean grain size – 4 μm, contents: non-combined C – 0.85%, O – 0.56%, N – 0.32%, $d\varepsilon/dt = 7 \times 10^{-5}$ s^{-1} [356, 634]; 16 – sintered $ZrC_{\sim 1.0}$ [319, 613]; 17 – $ZrC_{\sim 1.0}$ [614, 635]; 18 – hot-pressed $ZrC_{0.77–0.84}$, porosity – 1–5%, content non-combined C – 1.6–2.5% [356]; 19 – $ZrC_{\sim 1.0}$ [213]; 20 – hot-pressed $ZrC_{0.84\pm0.01}$, porosity – 2%, content non-combined C – 4.1–4.3% [350]; 21 – $ZrC_{\sim 1.0}$ [351] (**b**: *curves* 12a, 12b and 12c are corresponding to the various values of strain rate $d\varepsilon/dt$; **c**: reduction in area ψ is indicated for *curves* 7, 13 and 15, for all the curves $\psi = 0$ at 1600 °C)

continues to be brittle even at around 2700 °C [335]. The examples of tensile stress-strain graphic diagrams (curves) of ZrC_{1-x} materials for various temperatures, strain rates and microstructural characteristics (mean grain size) are given in Fig. 5.23. The variations of flexural (bending) strength σ_f with carbon content within the homogeneity range and temperature within the wide range from room to ultra-high temperatures are presented in Figs. 5.24 and 5.25. The effect of surface conditions on the flexural (bending) strength σ_f at room temperature for the materials with various porosity is demonstrated by data given in Table 5.14. The data on compressive strength σ_c of near-stoichiometric ZrC_{1-x} materials are summarized in Fig. 5.26. For the relationships between strength characteristics and microstructural factors (porosity $P \leq 30–35\%$ and mean grain size L, 3–24 μm) of ZrC_{1-x} materials Bulychev et al. [4, 86] proposed the following equations:

$$\sigma_f = 540(1 - 2.8P), \tag{5.78}$$

$$\sigma_f = 1490L^{-0.45}, \tag{5.79}$$

$$\sigma_c = 1830(1 - 3.3P), \tag{5.80}$$

while Kruglov et al. [610] for the flexural (bending) strength σ_f proposed the modified expression (for the structures with mean grain size L in the range from 8 to 60 μm)

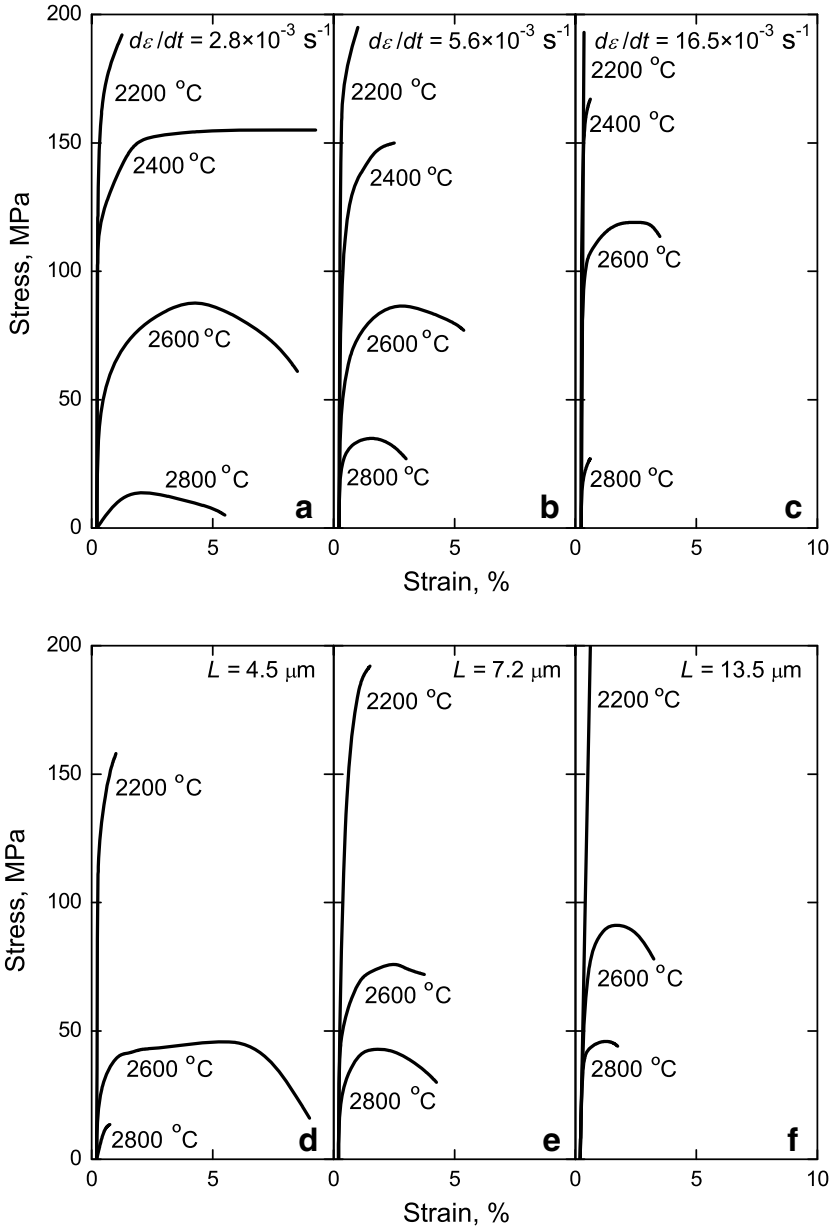

Fig. 5.23 The examples of tensile stress-strain graphic diagrams (curves) of sintered $ZrC_{\sim 1.0}$ materials with the various values of strain rates $d\varepsilon/dt$ (**a**, **b**, **c**) and mean grain sizes L (**d**, **e**, **f**) at high and ultra-high temperatures [601]

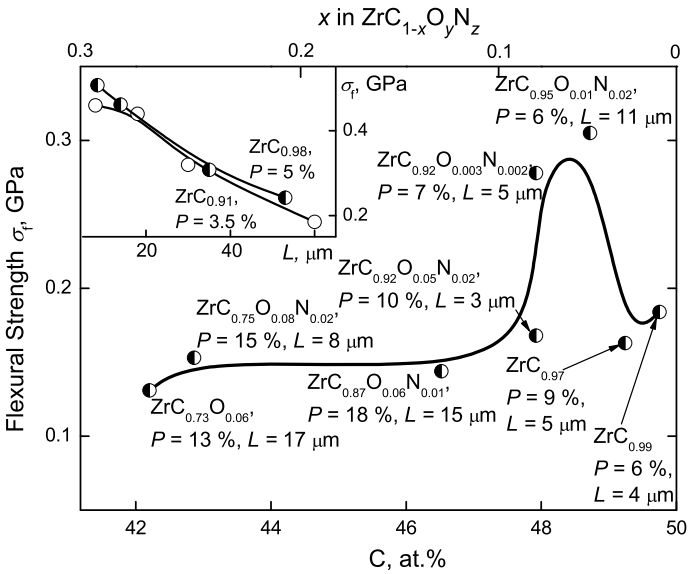

Fig. 5.24 Variation of the flexural (bending) strength σ_f at room temperature of sintered zirconium monocarbide $ZrC_{1-x}O_yN_z$ (contents of oxygen and nitrogen determined and marked; content non-combined carbon – 1.3% ($x = 0.01$), 2.2% ($x = 0.03$), 0.22% ($x = 0.05$), 0.15% ($x = 0.08$, $y = 0.003$), 0.05% ($x = 0.08$, $y = 0.05$; $x = 0.13$; $x = 0.25$; $x = 0.27$)) materials with deviation from the stoichiometry within the homogeneity range; porosity P and mean grain size L are indicated for all the compositions [606] (*Inset* – for sintered (in Ar) $ZrC_{0.91}$ (contents: non-combined C – 0.02%, O < 0.05%) [610] and sintered $ZrC_{0.98}$ [333], porosity indicated, in σ_f, GPa – L, μm scale)

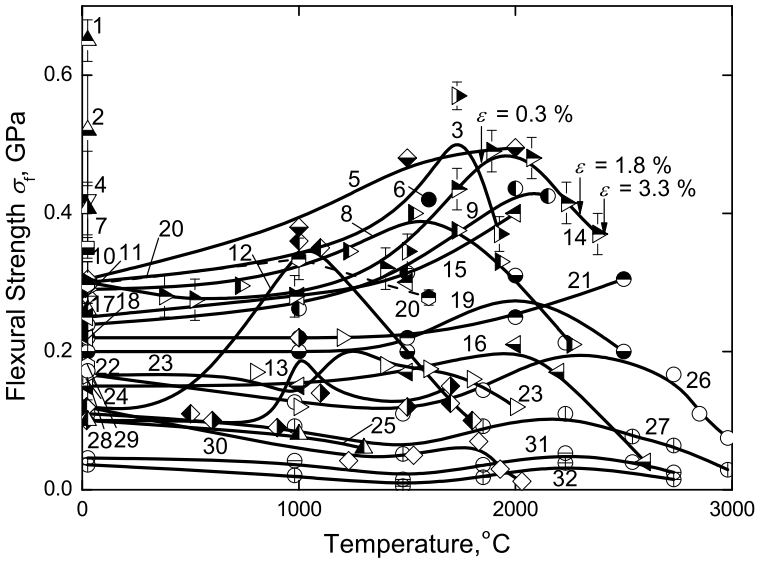

Fig. 5.25 (see caption on page 489)

◀**Fig. 5.25** Variation of flexural (bending) strength σ_f of near-stoichiometric zirconium monocarbide ZrC_{1-x} materials with temperature on the basis of several sources: 1 – sintered non-isothermally $ZrC_{0.95}$, porosity – 8%, mean grain size – 12–18 μm, contents: non-combined C – 0.04%, O – 0.33%, N – 0.03%, W – 0.40%, 3-point bending scheme [623]; 2 – sintered $ZrC_{0.96-1.00}$, porosity – 2–3%, mean grain size – 1–10 μm [86]; 3 – sintered $ZrC_{\sim 1.0}$, porosity – 5%, mean grain size – 6–16 μm, contents: non-combined C – 0.30%, O – 0.08%, N – 0.01%, W – 0.90%, 2.75 × 2.75 × 60 mm bar, 4-point bending scheme [625–627]; 4 – sintered $ZrC_{\sim 1.0}$, porosity – 3–7%, mean grain size – 3–14 μm [440, 441, 536, 541, 605]; 5 – pyrolytic $ZrC_{\sim 1.0}$ [363]; 6 – sintered (in Ar) $ZrC_{\sim 1.0}$, porosity – 10–12%, mean grain size – 1–2 μm [1, 612]; 7 – spark plasma sintered $ZrC_{\sim 1.0}$, porosity – 1–2%, mean grain size – 13 ± 1 μm, 0.8 × 1.0 × 10 mm bar, 3-point bending scheme [587, 588, 591]; 8 – sintered $ZrC_{\sim 1.0}$, porosity – 6%, mean grain size – 8 μm [330]; 9 – sintered $ZrC_{\sim 1.0}$, porosity – 5% [601]; 10 – sintered $ZrC_{\sim 1.0}$, porosity < 2%, mean grain size – 6 μm, 2 × 3 × 15 mm bar, 3-point bending scheme [631]; 11 – sintered $ZrC_{0.96}$, porosity – 6%, contents: O ≤ 0.05%, N – 0.05%, W – 0.2–0.3%, 2.5 × 2.5 × 30 mm bar, 3-point bending scheme [617]; 12 – sintered $ZrC_{0.95}$, porosity – 2.5%, mean grain size – 10 μm, thickness – 1 mm, 4-point bending scheme [537]; 13 – same, thickness – 2 mm, 4-point bending scheme [537]; 14 – sintered $ZrC_{\sim 1.0}$, porosity – 5%, $d\varepsilon/dt = 5 \times 10^{-3}$ s^{-1} [333, 336]; 15 – $ZrC_{0.96}$ obtained by metal right-through saturation (in Ar), mean grain size – 250 μm, 0.8 × 1.5 × 20 mm bar, 3-point bending scheme [611]; 16 – same, 0.6 × 2.0 × 20 mm bar, 3-point bending scheme [611]; 17 – sintered and annealed materials (Weibull modulus $m = 6.0$) [3]; 18 – sintered $ZrC_{0.93}$, porosity – 4–8%, mean grain size – 66 μm, contents: non-combined C – 0.5%, W – 0.2%, 3-point bending scheme (measured values at the temperature of liquid nitrogen proved to be practically equal to the corresponding values at 20 °C) [411, 624, 628, 687]; 19 – sintered $ZrC_{0.95}$, porosity – 6%, contents: O + N – 0.69%, $d\varepsilon/dt = 2.8 \times 10^{-3}$ s^{-1} [4, 645]; 20 – hot-pressed $ZrC_{\sim 1.0}$, porosity – 2%, content O – 0.36% [1479]; 21 – sintered $ZrC_{0.95}$, porosity – 6%, contents: O + N – 0.69%, $d\varepsilon/dt = 16.5 \times 10^{-3}$ s^{-1} [4, 645]; 22 – metal carbidized $ZrC_{0.96}$ [545]; 23 – hot-pressed $ZrC_{\sim 1.0}$, porosity – 4%, 6.3 × 12.7 × 76 mm bar, loaded in perpendicular direction to hot-pressing axis, 3-point bending scheme for room temperature and 4-point bending scheme for high temperatures, calculated strain at fracture at room temperature – 0.04% [308]; 24 – metal carbidized $ZrC_{\sim 1.0}$, mean grain size – 200–300 μm, contents: O + N < 0.56% [545]; 25 – sintered $ZrC_{0.98}$, porosity – 9%, mean grain size – 50 μm, 4-point bending scheme [537]; 26 – slip-cast and sintered $ZrC_{0.93}$, porosity – 14%, mean grain size – 16–18 μm, content non-combined C – 0.3% [603]; 27 – cold-pressed and sintered $ZrC_{0.93}$, porosity – 10%, content non-combined C – 0.3% [603]; 28 – sintered $ZrC_{0.97}$, 1.7 × 4.2 × 30 mm bar, 3-point bending scheme [618]; 29 – sintered and annealed (in Ar) $ZrC_{\sim 1.0}$, 0.8 × 2.0 × 13 mm bar, 3-point bending scheme [616]; 30 – hot-pressed (low-porous) materials [608, 609]; 31 – slip-cast and sintered $ZrC_{0.93}$, porosity – 40%, mean grain size – 8–10 μm, content non-combined C – 0.3% [603]; 32 – cold-pressed and sintered $ZrC_{0.93}$, porosity 37%, content non-combined C – 0.3% [603] (bending strain values ε are indicated for *curve* 14 at corresponding temperatures)

$$\sigma_f = K(1/L - 1/L_0)^{1/2}, \tag{5.81}$$

where K is a constant and L_0 is a critical grain size determined by the scale factor and stressed state. In the insets to Figs. 5.24 and 5.26 the effect of mean grain size magnitude of ZrC_{1-x} materials on the strength characteristics is demonstrated specially. The approximate general relationship $\sigma_t : \sigma_f : \sigma_c \approx 1 : (1.5–3) : (7–12)$ can be recommended for the common estimations of the strength characteristics of monocarbide ZrC_{1-x} materials [3, 33, 86, 330, 604, 619, 629]. According to Lanin et al. [411, 617, 660, 661], at room temperature sintered $ZrC_{0.93}$ (porosity – 4–8%, mean grain size – 66 μm, contents: non-combined C – 0.5%, W – 0.2%) has the fracture toughness $K_{IC} = 3.5$ MPa $m^{1/2}$ and sintered $ZrC_{0.96}$ (porosity 7%, mean grain size 20 μm, contents: non-combined C – 0.1%, O – 0.1%, N – 0.04%) – 2.8 MPa $m^{1/2}$ (with the ratio $K_{IA}/K_{IC} = 0.94$, scattered from 0.89 to 0.99); earlier for sintered $ZrC_{0.95}$ (porosity 8%) Warren [662] estimated K_{IC} to be in the range from 1.1 ± 0.1 to 2.3 ± 0.3 MPa $m^{1/2}$

Table 5.14 The effect of surface condition/treatment on the flexural (bending) strength at room temperature for sintered $ZrC_{0.90}$ materials with various porosity [602]

Porosity, %	Treatment character	Variations of ultimate flexural (bending) strength, MPa for the different efficiency of spark cutting treatment, 10^{-6} m^2 min^{-1}			
		0.8	1.6	5.0	18.0
3	SCT[a]	279	230	204	204
	SCT + BCP[b]	330	319	286	281
8	SCT	–	176	141	132
	SCT + BCP	–	225	221	222
18	SCT	–	118	114	106
	SCT + BCP	–	169	171	170

[a]SCT – spark cutting treatment
[b]SCT + BCP – spark cutting treatment with subsequent polishing by boron carbide powder

with the average fracture surface energy variation from 1.7 to 7.7 J m^{-2}. Katoh et al. [96] employing different methods for the K_{IC} determination of high-purity zone-refined $ZrC_{0.93}$ got 1.40 ± 0.02 MPa m$^{1/2}$ measured by Vickers indentation technique and 2.7 ± 0.3 MPa m$^{1/2}$ – by chevron-notched beam flexure method (ASTM C1421); Wang et al. [1425] by direct crack measurements – (1.9–3.0) ± (0.4–0.8) MPa m$^{1/2}$ for hot-pressed $ZrC_{0.90}$ (porosity – 1–2%, mean grain size – 50–100 μm). Fracture toughness K_{IC} of magnetron-sputtered nanocrystalline $ZrC_{0.80–1.0}$ films (thickness – 0.15–0.20 μm) was also reported in the range of 1.5–2.5 MPa m$^{1/2}$ [74, 663], and the value of K_{IC} for quasi-stoichiometric $ZrC_{\sim 1.0}$ recommended by Mukhopadhyay et al. [591] is around 2 MPa m$^{1/2}$. The fracture toughness K_{IC} of spark-plasma sintered $ZrC_{\sim 1.0}$ (porosity – 16%) is 2.5 ± 0.2 MPa m$^{1/2}$; for the same materials post-annealed K_{IC} = 3.1 ± 0.2 MPa m$^{1/2}$ [697]. For hot-pressed $ZrC_{\sim 1.0}$ materials (porosity – 2%) Zhao et al. [1479] determined K_{IC} = 2.62 ± 0.05 MPa m$^{1/2}$ by calculating indentation crack length. The highest values of fracture toughness were reported for reactive hot-pressed $ZrC_{0.96}$ (porosity < 2%, mean grain size – 5–10 μm, trace amount of non-combined C) K_{IC} =(4.0–4.7) ± 0.4 MPa m$^{1/2}$ (direct crack measurements) [1425]. The particularities of fatigue behaviour of ZrC_{1-x} materials were studied and analyzed in the works by Lanin with co-authors [333–335, 660, 661, 666, 667, 745].

The reported values for ductile-to-brittle transition temperatures of ZrC_{1-x} phases with various deviations from the stoichiometry are listed in Table 5.15. At the temperatures above ~1900–2000 °C yield strength $\sigma_{0.2}$ of ZrC_{1-x} is ruled by the Hall-Petch law [333]:

$$\sigma_{0.2} = \sigma_0 + k_y L^{-1/2}, \tag{5.82}$$

where σ_0 is the Peierl's stress, k_y is the coefficient, which characterizes the energy consumption for sliding transfer through the grain boundaries and L is the mean grain size; for sintered $ZrC_{0.98}$ (porosity – 3–4%, mean grain size – 5–8 μm, contents: non-combined C – 0.07%, O – 0.05%) under compression $\sigma_{0.2}$ falls off with temperature rise from 536 MPa at 1900 °C up to 278 MPa at 2200 °C (strain rate $d\varepsilon/dt$ = 10^{-2} s^{-1}) [742]. For practical applications of ultra-high temperature materials the values of prolonged strength (creep resistance) are of great importance. The examples of creep (deformation) kinetics of sintered and annealed zirconium

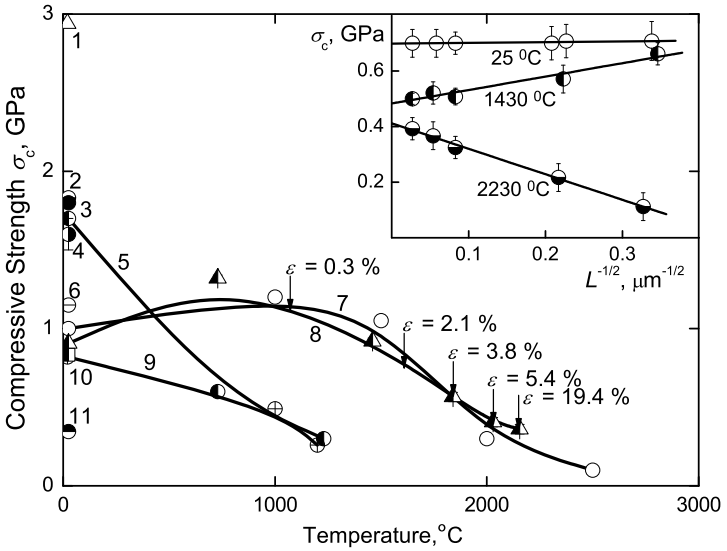

Fig. 5.26 Variation of compressive strength σ_c of near-stoichiometric zirconium monocarbide ZrC_{1-x} materials with temperature on the basis of several sources: 1 – $ZrC_{\sim 1.0}$, corrected to the poreless state [30, 213, 320, 635]; 2 – sintered $ZrC_{0.96}$, mean grain size – 3–10 μm, corrected to the poreless state [86]; 3 – sintered $ZrC_{0.97}$, porosity – 5%, mean grain size – 15–20 μm, $d\varepsilon/dt = 3.7 \times 10^{-3}$ s^{-1} [1]; 4 – sintered $ZrC_{\sim 1.0}$, porosity – 6%, mean grain size – 3 μm [37, 63, 319, 536, 567, 635, 636]; 5 – recommended for near-stoichiometric compositions [30, 33, 607]; 6 – $ZrC_{\sim 1.0}$ [3, 604]; 7 – sintered $ZrC_{0.95}$, porosity – 7%, mean grain size – 8 μm, contents: O + N – 0.14%, $d\varepsilon/dt = 3.0 \times 10^{-3}$ s^{-1} [4, 645]; 8 – sintered $ZrC_{\sim 1.0}$, $d\varepsilon/dt = 3.0 \times 10^{-3}$ s^{-1} [333, 336]; 9 – hot-pressed $ZrC_{\sim 1.0}$, porosity < 1% [1, 32, 608]; 10 – [1512]; 11 – pyrolytic $ZrC_{\sim 1.0}$ [363]; compression strain values ε are indicated for *curve* 8 at corresponding temperatures (*Inset* – for sintered $ZrC_{0.98}$, porosity – 5% [333, 621, 622], Hall-Petch relationships in σ_c, GPa – $L^{-½}$, μm$^{-½}$ scale)

Table 5.15 Ductile-to-brittle transition temperatures of zirconium monocarbide phases ZrC^{a}_{1-x}

Composition	Temperature, K (°C)	Characteristics	Reference
$ZrC_{\sim 1.0}$	2470 (2200)	Sintered, polycrystalline	[3]
$ZrC_{0.98}$	1420 (1150)	Sintered, polycrystalline[b]	[654–656]
$ZrC_{0.95}$	1370 (1100)	Single crystal[c]	[170]
$ZrC_{0.94}$	1470 (1200)	Arc-melted and annealed, polycrystalline[d]	[652]
$ZrC_{0.93}$	2270–2670 (2000–2400)	Sintered, polycrystalline[e]	[603]
$ZrC_{0.87–0.90}$	1170–1370 (900–1100)	Single crystal[f]	[657, 658]

[a]For near-stoichiometric ZrC_{1-x} the following average values were estimated: Peierls stress $\tau_P = 6.0$ GPa, shear modulus – Peierls stress ratio $\tau_P/G = 0.036$ and slip activation energy barrier $q = 1.6$ eV [52]
[b]Materials with mean grain sizes from 10 to 1000 μm, deformed under uniaxial compression
[c]Materials with contents: O – 0.005%, N – 0.067%
[d]Materials with content O – 0.04%, deformed under compression
[e]Materials with porosity – 14%, mean grain size – 16–18 μm, content non-combined C – 0.3%, deformed under flexure (bending) loading
[f]Compression testing

monocarbide ZrC_{1-x} materials are given in Fig. 5.27; the variations of long-term rupture strength σ_r for sintered $ZrC_{0.98}$ materials with time-to-rupture and temperature (1-, 10- and 100-h strength characteristics) are demonstrated in Fig. 5.28. Similar to some other ceramics and metals, the time-to-rupture τ_r of ZrC_{1-x} materials is related with applied stress σ exponentially [638]:

$$\tau_r = B\sigma^{-m}\exp(U/RT), \tag{5.83}$$

where T is temperature, K, R is the gas constant, U is the activation energy of long-term rupture, B and m are pre-exponential and exponent constants, respectively, and the effect of temperature on σ_r is well described by the Ito-Shishokin equation:

$$\sigma_r = C\exp(-\alpha T), \tag{5.84}$$

where C and α are constants. High- and ultra-high temperature steady-state creep rate $d\varepsilon/dt$ characteristics for the various compositions of ZrC_{1-x} materials are shown in Fig. 5.29. Creep rate variations of sintered ZrC_{1-x} materials with deviation from the stoichiometry within the homogeneity range are shown in Fig. 5.30. On the basis of all the previous research Frost and Ashby [659] plotted first stress–temperature (plasticity and creep deformation mechanism) and strain-rate–stress maps/diagrams for transition metal carbide materials, in particular for $ZrC_{0.95}$ of grain size 10 μm, followed later by Zubarev [638], who constructed similar map/diagram for $ZrC_{0.93}$ with mean grain size 15 μm (the diagrams are not given in the reference book); Fig. 5.31 has the special form of the detailed normalized shear stress – homologous temperature map/diagram for sintered $ZrC_{\sim 1.0}$ materials (mean grain size – 14 μm), which was constructed by Zubarev et al. [643, 644] on the basis of latest precise investigations. The formal creep characteristics (activation

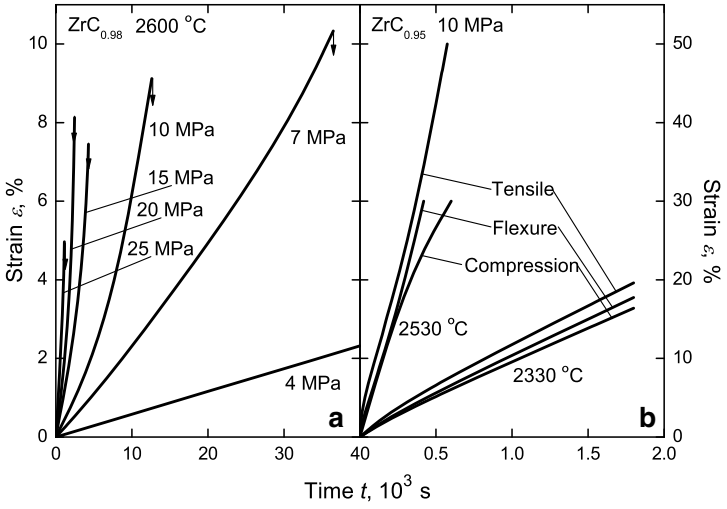

Fig. 5.27 High- and ultra-high temperature creep of sintered and annealed zirconium monocarbide ZrC_{1-x} materials: (**a**) tensile characteristics of $ZrC_{0.98}$ (porosity – 3%, mean grain size – 15 μm, contents: non-combined C – 0.40%, O – 0.07%, N – 0.025%) at 2600 °C and 4–25 MPa and (**b**) tensile, flexure (bending) and compession characteristics of $ZrC_{0.95}$ (porosity – 15–17%, mean grain size – 4.5 μm) at 2330–2530 °C and 10 MPa [637, 638]

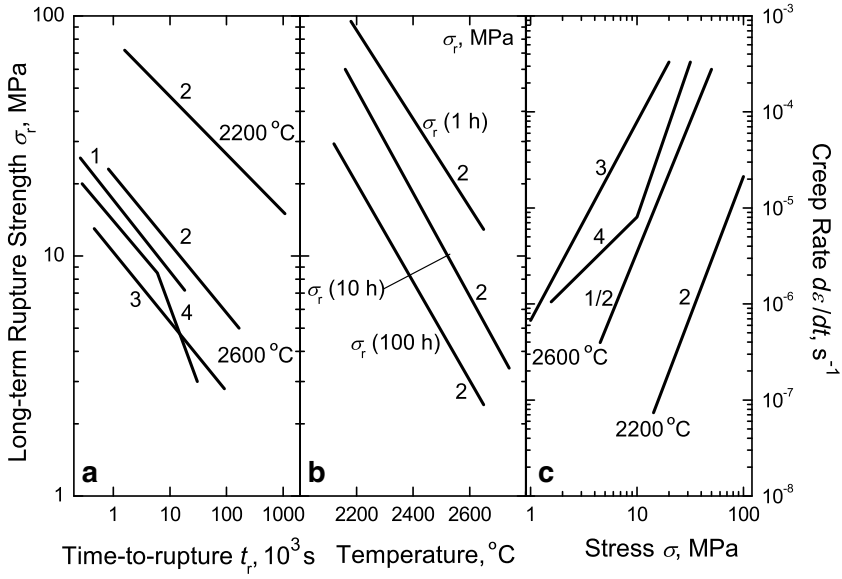

Fig. 5.28 Variations of long-term rupture strength σ_r with time-to-rupture (**a**) and temperature (for 1-, 10- and 100-hour strength characteristics) (**b**), and variation of steady-state creep rate $d\varepsilon/dt$ with applied stress σ (**c**) of sintered and annealed $ZrC_{0.98}$ materials with various compositions and structures: 1 – contents: non-combined C – 0.35%, O – 0.08%, N – 0.05%, W – 0.21% (porosity – 4%, mean grain size – 15 μm); 2 – contents: non-combined C – 0.40%, O – 0.07%, N – 0.025% (porosity – 3%, mean grain size – 15–20 μm); 3 – contents: non-combined C – 1.0%, O – 0.08%, N – 0.04%, W – 0.21% (porosity – 4–5%, mean grain size – 5–6 μm); 4 – contents: non-combined C – 1.1%, O – 0.10%, N – 0.04% (porosity – 3–4%, mean grain size – 15–20 μm) [638–640]

energy Q, exponent constant n) of zirconium monocarbide phases ZrC_{1-x} at various temperature and stress ranges are listed in Table 5.16. Approximately, the following main dominating creep mechanisms can be established in some overlapping parametric ranges for zirconium monocarbide ZrC_{1-x} materials with certain changes within the homogeneity range [638–644, 646, 668, 669]:

vacancy-diffusional (or Nabarro-Hering) mechanism controlled by the bulk diffusion of metal in carbide (at temperatures from 2400 to 3000 °C and applied stresses from 5 to 60 MPa, with $Q \approx 660$–820 kJ mol^{-1} and $n \approx 1.0$–1.9);
dislocation-diffusional mechanism controlled by the bulk diffusion of metal in carbide (at 2200–3000 °C and 30–100 MPa, with $Q \approx 660$–840 kJ mol^{-1} and $n \approx 3$–8);
viscous dislocation sliding mechanism controlled by the bulk diffusion of carbon in carbide (at 1800–2200 °C and 60–280 MPa, with $Q \approx 370$–470 kJ mol^{-1} and $n \approx 3.0$–4.5).

Some other plasticity and creep mechanisms of sintered $ZrC_{\sim 1.0}$ and their parametric ranges, including the comparison with single crystal materials, are indicated in the graphic scheme and legend of Fig. 5.31. Darolia and Archbold have found out that arc-melted and annealed $ZrC_{0.94}$ materials can undergo dynamic recrystallization at 1800 °C during the creep compression testing with strain rate of $\sim 3\times10^{-4}$ s^{-1} [1310]. The particularities of the irradiation induced creep of ZrC_{1-x} materials were considered by Katoh et al. [757].

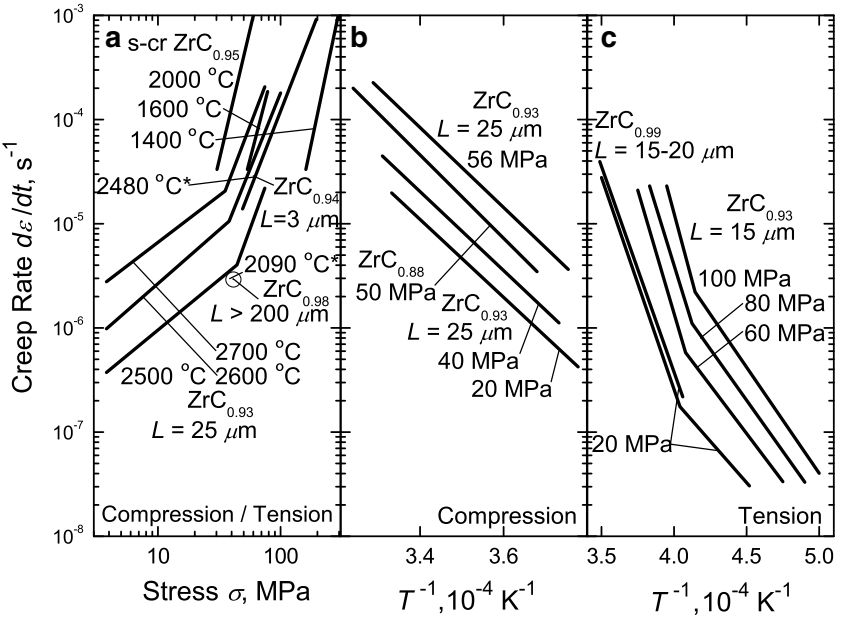

Fig. 5.29 High- and ultra-high temperature steady-state creep rate $d\varepsilon/dt$ variations of single crystal $ZrC_{0.95}$ (contents: O – 0.005%, N – 0.067%) at 1400–2000 °C [170], arc-fused and cast $ZrC_{0.98}$ (mean grain size > 200 μm) at 2090 °C* [320, 653], sintered $ZrC_{0.99}$ (porosity – 3%, mean grain size – 15–20 μm, contents: non-combined C – 0.38%, O – 0.06%, N – 0.025%, Fe – 0.09%) at 20 MPa [639, 640], sintered and annealed $ZrC_{0.94}$ (porosity – 15–20%, mean grain size – 3 μm, contents: non-combined C – 0.17%, O – 0.17%, N – 0.03%, Fe – 0.11%) at 2480 °C* [639, 640], sintered $ZrC_{0.93}$ (porosity – 2–4%, mean grain size – 15 μm and 25 μm) at 2500–2700 °C and 20–100 MPa [638, 641, 642] and sintered $ZrC_{0.88}$ materials at 50 MPa [119]: (**a**) with applied stress under compression/tension (tension temperature marked with asterisk), (**b**) with reciprocal temperature under compression and (**c**) with reciprocal temperature under tension

At room temperature the values of Young's E, Coulomb's (shear) G and bulk (compression) K moduli, volume compressibility κ and Poisson's ratio ν of sub- and near-stoichiometric zirconium monocarbide ZrC_{1-x} materials lie within the areas: E = 300–500 GPa [1, 3, 4, 11, 26, 27, 30, 32, 33, 39, 42, 46–49, 52, 54, 74, 82, 96, 101, 452, 466, 677, 678, 684, 692, 706], G = 120–200 GPa [3, 4, 11, 39, 49, 69, 82, 125, 132, 277, 284, 466, 659, 675–678, 703], K = 170–340 GPa [3, 4, 11, 49, 82, 125, 466, 675–678, 684, 706], κ = 4.25–4.75 TPa^{-1} [1, 3, 4, 54, 725, 1372, 1514] and ν = 0.16–0.26 [3, 4, 11, 39, 49, 82, 123, 132, 277, 308, 670, 677, 678, 684, 686, 689, 696, 706] (*see also* Table 5.17); the experimentally measured velocity of ultrasonic waves propagated in quasi-stoichiometric zirconium monocarbide $ZrC_{\sim 1.0}$ (summarized on the basis of several sources) [736]:

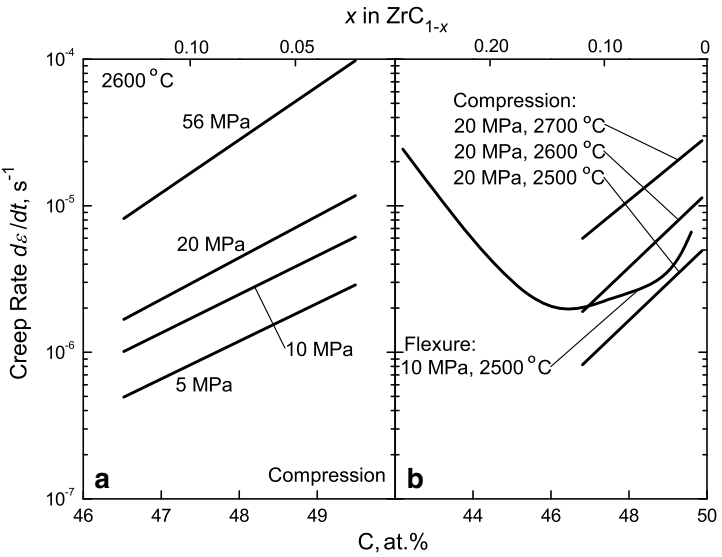

Fig. 5.30 Creep rate $d\varepsilon/dt$ variations of zirconium monocarbide ZrC_{1-x} materials with deviation from the stoichiometry within the homogeneity range: (**a**, **b**) under compession (sintered materials, porosity – 2–5%, mean grain size – 25 μm [118, 638]) and (**b**) under flexure (hot-pressed materials, porosity – 4–6%, mean grain size – 20–70 μm, non-combined C – 0.10%, O – 0.03–0.47%, N – 0.01–0.23% [599]) loading (temperatures and applied stresses are marked)

average velocity V_m, m s^{-1} 6270,

and theoretically calculated sound velocities for near-stoichiometric zirconium monocarbide $ZrC_{\sim 1.0}$ are

by *ab initio* and similar theoretical calculations (for zero pressure):

longitudinal velocity V_S, m s^{-1} 13560 [701], 8850 [725, 1372], 8194 [24], 8641 (0 K) [283], 7981 (~300 K) [283], 7984 [376];

transversal velocity V_T, m s^{-1} 4500 [701], 5280 [725], 4968 [24], 5254 (0 K) [283], 4662 (~300 K) [283], 4788 [376];

average velocity V_m, m s^{-1} 5129 [701], 5840 [725], 5481 [24], 5297 [376];

by LMTO (linearized method of "muffin-tin" orbitals) calculations [736]:

average velocity V_m, m s^{-1} 5970;

by TBFP (three body force potential) modeling calculations [725]:

longitudinal velocity V_S, m s^{-1} 8170;

transversal velocity V_T, m s^{-1} 5020;

average velocity V_m, m s^{-1} 5540.

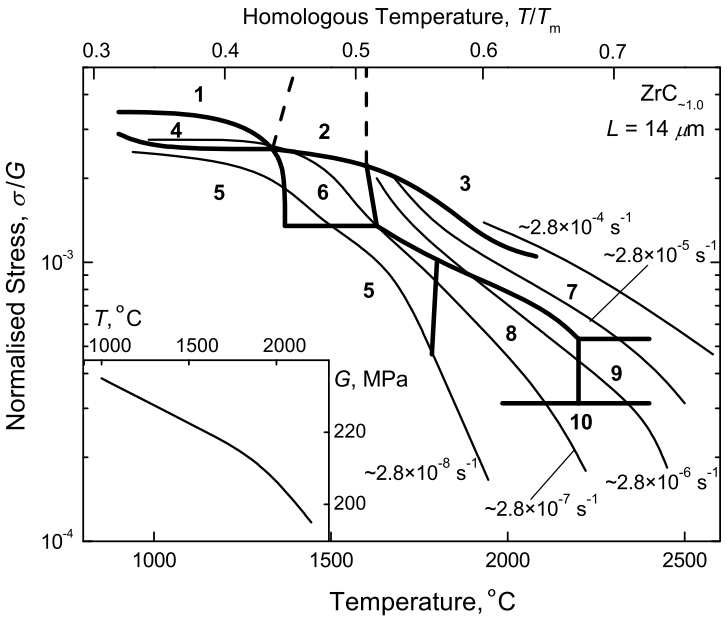

Fig. 5.31 A normalized shear stress σ/G – homologous (to melting point) temperature T/T_m map (creep mechanism diagram) for quasi-stoichiometric sintered $ZrC_{\sim 1.0}$ materials (mean grain size – 14 μm, contents: non-combined C – 0.07%, O – 0.06%, N – 0.03%) under compression loading (2.7 × 2.7 × 8.0 mm bar) in He atmosphere, demarcated areas with different creep mechanisms: 1 – quasi-brittle fracture (< ~ 1400 °C); 2 – applied stress exceeds the yield strength (~ 1350–1650 °C, creep characteristic: activation energy Q = 210–435 kJ mol^{-1}, maximal values of yield strength); 3 – applied stress exceeds the yield strength (> ~ 1650 °C, Q = 210–435 kJ mol^{-1}, highest creep rates); 4 – interaction with forest dislocations (presence of long-term non-steady-state creep stages with $\varepsilon \approx 0.1$, < ~ 1350–1400 °C, Q = ~ 15 kJ mol^{-1}); 5 – generation of dislocations (< ~ 1800 °C, Q = 480–1270 kJ mol^{-1}); 6 – cross-gliding of dislocations (~ 1350–1650 °C, >200 MPa, creep characteristics for polycrystalline materials with grain size 14–1000 μm: Q = 270–365 kJ mol^{-1}, exponent constant n = 4–7; creep characteristics for single crystal materials: Q = 405 kJ mol^{-1}); 7 – climb of dislocations by means of bulk (lattice) diffusion (dislocation-diffusional mechanism, > 1600 °C, polycrystalline materials: Q = 815–835 kJ mol^{-1}, n = 8.0; single crystal: Q = 930 kJ mol^{-1}, n = 5.8–6.3); 8 – climb of dislocations by means of diffusion through dislocation core (nucleus) or Burgers (screw) dislocation gliding (polycrystalline materials: Q = 440–470 kJ mol^{-1}, n = 4.2–4.4; single crystal: Q = 365 kJ mol^{-1}, n = 1.6); 9 – grain boundary sliding (>2200 °C, Q = 795–835 kJ mol^{-1}, n = 4.0–4.3); 10 – diffusion mass transfer (Nabarro-Herring (vacancy-diffusional) creep mechanism, Q = 795–815 kJ mol^{-1}, n = 1.4–1.9); for some areas the comparison of characteristics with single crystal $ZrC_{0.94}$ under compession loading (2.0 × 2.0 × 6.0 mm bar (100) faced) along <100> direction is given above, isolines of creep rate $d\varepsilon/dt$ from ~ 2.8 × 10^{-4} s^{-1} up to ~ 2.8 × 10^{-8} s^{-1} are indicated; (*Inset* – variation of Coulomb's (shear) modulus G of $ZrC_{\sim 1.0}$ with temperature) [643, 644]

Table 5.16 Formal creep characteristics (activation energy Q, stress exponent constant n) of zirconium monocarbide phases ZrC_{1-x} at various temperature and stress ranges

Composition	Load type[a]	Temperature range, °C	Stress range[b], MPa	Activation energy[b], Q, kJ mol^{-1}	Exponent constant[b], n	Reference
$ZrC^{c}_{\sim 1.0}$	C	1350–1400	450–900	15	–	[643, 644]
	C	1000–1800	200–600	480–1270	–	[643, 644]
	C	1350–1650	280–680	270–365	4.0–7.0	[643, 644]
	C	1350–1650	680–1100	210–435	–	[643, 644]
	C	1650–2100	200–680	210–435	–	[643, 644]
	C	1600–2400	90–680	815–835	8.0	[643, 644]
	C	2200–2400	50–90	795–835	4.0–4.3	[643, 644]
	C	1800–2200	70–200	440–470	4.2–4.4	[643, 644]
	C	2000–2400	20–70	795–815	1.4–1.9	[643, 644]
$ZrC_{\sim 1.0}$	C	1700–2000	–	580	–	[646, 650]
$ZrC^{d}_{\sim 1.0}$	H	700–1660	–	45	–	[573, 574]
	H	1660–2650	–	200	–	[573, 574]
$ZrC^{e}_{\sim 1.0}$	R	~1700	–	95	3.2	[335]
$ZrC^{f}_{0.99}$	T	2200–2750	2–75	710 ± 40	2.3–3.1	[639, 640]
$ZrC^{g}_{0.99}$	T	2600	2–10	–	1	[639, 640]
$ZrC^{h}_{0.98}$	F	2400–2700	0–15	760 ± 50	1	[3, 4, 599]
$ZrC_{0.98}$	T,F,C	1800–3000	$<\sigma_{lc}$	660	1	[118, 647]
	F	<2400	$>\sigma_{lc}$	370–440	2.8–3.2	[118, 647]
$ZrC^{i}_{0.98}$	F	1800–2200	120–280	370	3.4	[638]
$ZrC^{j}_{0.97}$	F	1800–2300	120–210	335	3	[642, 646]
$ZrC^{k}_{0.97}$	C	2200–2600	6.5–55	690 ± 60	1	[4, 649]
$ZrC^{l}_{0.97}$	C	2200–2600	6.5–55	540 ± 40	1	[4, 649]
$ZrC^{m}_{0.96}$	F	2400–2700	0–15	700 ± 40	1	[3, 4, 599]
$ZrC^{n}_{0.96}$	C	2500–2700	5–20	660	1	[4, 638]
$ZrC_{0.96}$	F	>2400	$>\sigma_{lc}$	660–720	3.4–3.7	[118, 647]
$ZrC^{o}_{0.95}$	–	1400–2000	40–270	460	5	[170, 707]
$ZrC^{p}_{0.95}$	T	2180–2540	3.5–20	485 ± 75	1	[646, 651]
$ZrC^{q}_{0.94–0.95}$	T,F,C	2150–2630	2–20	520 ± 40	1.0–1.1	[118, 637]
$ZrC^{r}_{0.94}$	C	1800–2000	140–200	405	–	[643, 644]
	C	2000–2200	20–140	365	1.6	[643, 644]
	C	2200–2400	20–140	930	5.8–6.3	[643, 644]
$ZrC^{s}_{0.94}$	C	2200–2600	3–60	680	1	[638]
$ZrC^{t}_{0.94}$	C	2500–2750	1–30	710 ± 40	1	[638, 646]
	C	2500–2750	30–75	710 ± 40	2.3–3.7	[638, 646]
$ZrC^{u}_{0.94}$	F	1800–2200	70–240	385	3.3	[638]
$ZrC^{v}_{0.94}$	C	1200–1800	–	500 ± 20	3.8–6.5	[646, 652]
$ZrC_{0.94}O^{w}_{0.05}$	C	1500–1600	60–100	–	0.7 ± 0.4	[664, 665]
	C	1500–1600	100–140	690 ± 40	$\geq$3.3 ± 0.4	[664, 665]

(continued)

Table 5.16 (continued)

Composition	Load type[a]	Temperature range, °C	Stress range[b], MPa	Activation energy[b], Q, kJ mol^{-1}	Exponent constant[b], n	Reference
$ZrC^{x}_{0.93\pm0.01}$	T	1800–2250	2.5–35	315	3	[356, 634]
	T	2150–2500	2.5–35	840	3	[356, 634]
$ZrC^{y}_{0.93}$	T	2200–2600	20–60	680	1	[638]
	C	2400–2700	5–50	700	1	[638, 641]
	T	2200–2400	60–100	720	3.1	[638]
	C	2500–2700	50–80	700	3.4	[638, 641]
	T	1800–2200	60–120	400	3.2	[638]
$ZrC^{z}_{0.93}$	C	2400–2700	5–40	680–695	1	[4, 638]
	C	2500–2700	35–80	680	3.1	[638]
$ZrC^{a1}_{0.93}$	C	2200–2700	5–40	690 ± 40	1	[4, 638]
	C	2200–2700	40–690	690 ± 40	3.4	[4, 638]
$ZrC_{0.93}$	F	2600	$>\sigma_{lc}$	660–720	3.4–3.7	[118, 647]
$ZrC^{l}_{0.92}$	T	2200–3000	3–80	750	2.5	[3, 642]
$ZrC^{a2}_{0.92}$	C	2200–2700	5–45	660	1	[638, 646]
	C	2200–2700	55–70	660	3.4	[638, 646]
$ZrC^{r}_{0.92}$	F	1900–2200	80–240	440	3.2	[638]
$ZrC_{0.92}$	F	>2400	$>\sigma_{lc}$	660–720	3.4–3.7	[118, 647]
	F	<2400	$>\sigma_{lc}$	370–440	2.8–3.2	[118, 647]
$ZrC^{a3}_{0.90}$	H	900–1500	–	310–330	3.8–4.1	[93, 172]
$ZrC^{n}_{0.89}$	C	2500–2700	5–20	740	1	[4, 638]
$ZrC_{0.89}$	T,F,C	1800–3000	$<\sigma_{lc}$	740	1	[118, 647]
$ZrC^{a4}_{0.88}$	F	2400–2700	0–15	730 ± 40	1	[3, 4, 599]
$ZrC^{a5}_{0.88}$	C	2450–3100	10–70	770	–	[119, 648]
$ZrC^{a6}_{0.85}$	F	2400–2700	0–15	680 ± 40	1	[3, 4, 599]
$ZrC^{a7}_{0.81}$	F	2400–2700	0–15	660 ± 40	1	[3, 4, 599]
$ZrC_{0.79}O^{a8}_{0.13}$	C	1500–1600	60–100	–	0.8 ± 0.4	[664, 665]
	C	1500–1600	100–140	–	≥2.7 ± 0.4	[664, 665]
	C	1500–1600	100	740 ± 30	–	[664, 665]
	C	1500–1600	140	480 ± 30	–	[664, 665]
$ZrC^{a9}_{0.73}$	F	2400–2700	0–15	580 ± 30	1	[3, 4, 599]
$ZrC^{b1}_{0.72}$	F	2400–2700	0–15	510 ± 30	1	[3, 4, 599]
$ZrC^{b2}_{0.63-0.90}$	F	–	–	380–690 (Q = 801–1143x)	–	[26, 27, 1277]

[a]Denoted: T – tension, F – flexure (bending), C – compression, H – hot hardness (measurement), R – relaxation

[b]$\sigma_{lc} = AL^{-0.388}$, where σ_{lc} is the ultimate linear stress, L is the mean grain size and A is a constant [4, 118, 647]; for Q and n *see* the Eq. (I-3.17) and consideration in [707]

[c]Sintered, mean grain size – from 14 to 1000 μm, contents: non-combined C – 0.07%, O – 0.06%, N – 0.03%

[d]Hot-pressed and annealed, porosity – 1–2%, mean grain size – ~50 μm, content non-combined C – 0.54%

[e]Numerical calculation of residual stresses for the relaxation rate in the complex field of thermal stresses

[f]Sintered, porosity – 3%, mean grain size – 15–20 μm, contents: non-combined C – 0.38–0.40%, O – 0.06–0.07%, N – 0.025%, Fe – 0.09%

[g]Sintered, porosity – 3–4%, mean grain size – 15–20 μm, contents: non-combined C – 1.1%, O – 0.10%, N – 0.04%

(continued)

Table 5.16 (continued)

[h]Hot-pressed, porosity – 4–6%, mean grain size – 30 μm, contents: non-combined C – 0.10%, O – 0.03%, N – 0.01%
[i]Sintered, porosity – 3–4%, mean grain size – 30 μm
[j]Sintered materials
[k]Sintered, porosity – 0.3–5.0%, mean grain size – 5–7 μm, contents: O + N – 0.15%
[l]Sintered, porosity – 19%, mean grain size – 7 μm, contents: O + N – 0.15%
[m]Hot-pressed, porosity – 4–6%, mean grain size – 8–17 μm, contents: non-combined C – 0.10%, O – 0.05%, N – 0.01%
[n]Sintered, porosity – 3–4%, mean grain size – 25 μm, contents: non-combined C – 0.25%, O + N – 0.07–0.14%
[o]Single crystal (direction – <111>), contents: O – 0.005%, N – 0.067%
[p]Sintered, porosity – 7%, mean grain size – 3–5 μm, contents: non-combined C – 0.25%, O – 0.12%, N – 0.03%, Fe – 0.04%
[q]Sintered, porosity – 15–17%, mean grain size – 4.5 μm, contents: non-combined C – 0.17%, O – 0.08%, N – 0.02%
[r]Single crystal (plane/direction – (100) <100>), contents: non-combined C – 0.05%, O – 0.10%, N – 0.02%
[s]Sintered, porosity – 3–4%, mean grain size – 6 μm
[t]Sintered, porosity – 3–4%, mean grain size – 5–45 μm
[u]Sintered, porosity – 3–4%, mean grain size – 14 μm
[v]Arc-melted, mean grain size – 250 μm, content O – 0.04%
[w]Spark plasma sintered, fully dense, mean grain size – 5–20 μm
[x]Hot-pressed, porosity – 7%, mean grain size – 4 μm, contents: non-combined C – 0.85%, O – 0.56%, N – 0.32%, Hf – 1.0%
[y]Sintered, porosity – 3–4%, mean grain size – 15 μm
[z]Sintered, porosity – 2–3%, mean grain size – 25 μm, contents: O + N – 0.12%
[a1]Sintered, porosity – 5%, mean grain size – 20 μm
[a2]Sintered, porosity – 2–4%, mean grain size – 6–45 μm
[a3]Single crystal (plane/direction – (100) <001> , (110) <001> and (111) <110>)
[a4]Hot-pressed, porosity – 4–6%, mean grain size – 6–65 μm, contents: non-combined C – 0.10%, O – 0.05%, N – 0.09%
[a5]Sintered, porosity – 4–9%, mean grain size – 23 μm
[a6]Hot-pressed, porosity – 4–6%, mean grain size – 16 μm, contents: non-combined C – 0.10%, O – 0.47%, N – 0.23%
[a7]Hot-pressed, porosity – 4–6%, mean grain size – 20 μm, contents: non-combined C – 0.10%, O – 0.29%, N – 0.06%
[a8]Spark plasma sintered, fully dense, mean grain size – 5 μm
[a9]Hot-pressed, porosity – 4–6%, mean grain size – 70 μm, contents: non-combined C – 0.10%, O – 0.29%, N – 0.06%
[b1]Hot-pressed, porosity – 4–6%, mean grain size – 45 μm, contents: non-combined C – 0.10%, O – 0.08%, N – 0.03%
[b2]For sintered ZrC_{1-x} ($0.10 \leq x \leq 0.37$) materials Q is described by the given equation

Table 5.17 Elastic properties (stiffness coefficients c_{11}, c_{12} and c_{44}, Young's modulus E, bulk (compression) modulus K, Poisson's ratio ν)[a] of zirconium monocarbide phases ZrC_{1-x}

Composition	Stiffness coefficients c_{ij}			Young's modulus E, GPa	Bulk modulus K, GPa	Poisson's ratio ν	Reference
	c_{11}, GPa	c_{12}, GPa	c_{44}, GPa				
$ZrC_{0.66}$[b]	–	–	–	266[c]	166[c]	0.23	[696]
$ZrC_{0.75}$	–	–	–	342[d,e]	200[d,e]	0.22[d]	[125, 675, 676]
	403[b]	105[b]	105[b]	301[b,f]	204[b,f]	–	[735]
$ZrC_{0.77}$	–	–	–	282[g]	–	0.21[g]	[123, 670, 689]
	–	–	–	353 ± 1[h]	–	–	[356, 634]
$ZrC_{0.78}$[i]	–	–	–	300[j]	173[j]	0.21	[4, 132]
$ZrC_{0.79}$	–	–	–	305[k,l]	147[k,l]	0.16[k]	[685]
	–	–	–	330[b,m]	187[b,m]	0.21[b]	[696]
$ZrC_{0.80}$	–	–	–	301[n]	–	0.20[n]	[123, 670, 689]
	–	–	–	–	225[o]	–	[276]
	–	–	–	–	188[p]	–	[699]
$ZrC_{0.83}$[b]	–	–	–	347[q]	196[q]	0.21	[696]
$ZrC_{0.84}$	–	–	–	353 ± 1[h]	–	–	[356, 634]
	–	–	–	346[r,s]	174[r,s]	0.17[r]	[685]
$ZrC_{0.85}$	–	–	–	369[d,t]	265[d,t]	–	[125, 675, 676]
	–	–	–	347[u]	–	0.20[u]	[123, 670, 689]
$ZrC_{0.87}$	–	–	–	365[v,w]	203[v,w]	0.20[v]	[4, 132]
	–	–	–	402 ± 4[x]	–	–	[96, 681]
	–	–	–	355[b,y]	203[b,y]	0.21[b]	[696]
$ZrC_{0.88}$[b]	460	104	139	374[z]	223[z]	–	[735]
$ZrC_{0.89}$[a1]	468	100	157	–	223	–	[49, 705, 706]
$ZrC_{0.90}$	–	–	–	387[a2,a3]	258[a2,a3]	0.25[a2]	[277]
	–	–	–	–	195[p]	–	[699]
$ZrC_{0.91}$	–	–	–	385[d,a4]	312[d,a4]	–	[125, 675, 676]
	–	–	–	382[a5]	–	–	[679]
$ZrC_{0.92}$	–	–	–	385[a6,a7]	225[a6,a7]	0.22[a6]	[4, 132]
	–	–	–	403[a8]	–	–	[575]
$ZrC_{0.93}$	–	–	–	390[a9]	–	–	[624]
	–	–	–	328[b1]	–	0.24[b1]	[411]
	–	–	–	320[a9]	–	0.20[a9]	[741]
$ZrC_{0.94}$[a1]	472	99	159	407[b2]	223[b2]	0.19	[11, 49, 684, 706]
$ZrC_{0.95}$	–	–	–	400[d,b3]	339[d,b3]	–	[125, 675, 676]
	–	–	–	400[b4,b5]	–	0.20[b4]	[96, 662]
$ZrC_{0.96}$	–	–	–	386[b6,b7]	207[b6,b7]	0.19[b6]	[11, 82]
	–	–	–	393[b8,b9]	240[b8,b9]	0.23[b8]	[4, 132]
	–	–	–	390[c1]	272[c1]	0.23[c1]	[123, 670, 689]
	–	–	–	320[c2]	–	–	[617]
	–	–	–	–	275[o]	–	[276]

(continued)

Table 5.17 (continued)

Composition	Stiffness coefficients c_{ij}			Young's modulus E, GPa	Bulk modulus K, GPa	Poisson's ratio ν	Reference
	c_{11}, GPa	c_{12}, GPa	c_{44}, GPa				
$ZrC^{b}_{0.97}$	484	108	146	394[c3]	234[c3]	–	[735]
$ZrC_{0.98}$	–	–	–	452[x,c4]	270[x,c4]	0.22[x]	[277]
	–	–	–	–	275[c5]	–	[276]
	–	–	–	394[c6,c7]	222[c6,c7]	–	[703]
$ZrC_{\sim 1.0}$	428[c8]	41[c8]	146[c8]	369[c9]	168[c9]	0.13[c9]	[49, 82, 677, 678]
	–	–	–	495[d1]	–	–	[3]
	–	–	–	410 ± 3[d2,d3]	–	0.17[d3]	[308]
	–	–	–	475[d4]	–	–	[673]
	–	–	–	392[d1]	–	–	[680]
	522[d5]	110[d5]	160[d5]	–	247[d5,d6]	–	[682]
	470[d7]	100[d7]	160[d7]	–	223[d7]	–	[683, 713]
	–	–	–	375[d8]	–	–	[576]
	–	–	–	345 ± 7	–	–	[246]
	–	–	–	310[d9]	–	–	[41]
	–	–	–	475	–	0.17	[686]
	–	–	–	394[e1]	–	–	[117]
	–	–	–	–	216[e2]	–	[688]
	–	–	–	–	218[e3]	–	[688]
	–	–	–	–	230[e4]	–	[690]
	–	–	–	350	–	–	[691, 692]
	–	–	–	348[d1]	–	–	[693, 1515]
	–	–	–	318[e5,e6]	214[e5,e6]	0.26[e5]	[39]
	–	–	–	388	–	–	[694]
	–	–	–	400[e7]	–	–	[628]
	–	–	–	–	237[e8]	–	[695]
	496[b]	109[b]	151[b]	408[b,e9]	238[b,e9,f1]	0.21[b]	[696]
	–	–	–	222[f2]	–	–	[697]
	–	–	–	–	221[p,f3]	–	[699]
	272[e8]	134[e8]	429[e8]	–	232[e8]	–	[68]
	–	–	–	506[e9,f4,f5]	–	0.14[e9,f4,f5]	[68]
	–	–	–	473[e8,f6,f7]	–	0.16[e8,f6,f7]	[68]
	452[b]	102[b]	154[b]	390[b,f8]	219[b,f8]	0.20[b]	[700]
	428[e8]	118[e8]	119[e8]	332[e8,f9]	222[e8,f9]	0.25[e8]	[701]
	477[g1]	98[g1]	162[g1]	–	–	–	[705]
	472[g2]	105[g2]	135[g2]	–	–	–	[705]
	480[g3]	102[g3]	170[g3]	–	–	–	[705]
	381[g4]	121[g4]	178[g4]	–	–	–	[167]

(continued)

Table 5.17 (continued)

Composition	Stiffness coefficients c_{ij}			Young's modulus E, GPa	Bulk modulus K, GPa	Poisson's ratio ν	Reference
	c_{11}, GPa	c_{12}, GPa	c_{44}, GPa				
	441	60	151	384[g5]	187[g5]	0.16	[708, 1369]
	–	–	–	216–325[g6]	–	–	[663]
	–	–	–	–	222[d5]	–	[709, 713]
	–	–	–	–	227[g7]	–	[710]
	–	–	–	–	214[g8]	–	[710]
	643[g2]	95[g2]	128[g2]	–	239[g2]	–	[711]
	512[g3]	73[g3]	121[g3]	–	–	–	[711]
	–	–	–	–	248[g2]	–	[712]
	–	–	–	–	236[g9]	–	[712]
	–	–	–	–	222[b]	–	[712]
	–	–	–	–	250[h1]	–	[714]
	–	–	–	312[h2,h3]	217[h2,h3]	0.19[h2]	[280]
	–	–	–	390–445[h4,h4]	265[h4,h5]	0.19[h4]	[65, 712]
	457[g3]	115[g3]	150[g3]	–	278[g3]	–	[715]
	–	–	–	464 ± 22[h6]	–	–	[588]
	548[h7]	87[h7]	87[h7]	525[h7,h8]	265[h7,h8]	0.13[h7]	[18, 716]
	504[g2]	90[g2]	173[g2]	438[f4,g2,h9,i1]	228[g2,i1]	0.18[f4,g2,i2]	[69]
	504[g2]	90[g2]	173[g2]	436[f6,g2,h9,i3]	228[g2,i3]	0.18[f6,g2,i3]	[69]
	499[g3]	93[g3]	170[g3]	436[f4,g4,h9,i4]	225[g3,i4]	0.18[f4,g3,i2]	[69]
	499[g3]	93[g3]	170[g3]	434[f6,g3,h9,i5]	225[g3,i5]	–	[69]
	455[i6]	116[i6]	152[i6]	408[i6,i7]	229[i6,i7]	0.21[i6]	[717]
	471[i8]	91[i8]	156[i8]	–	218[i8]	–	[281]
	–	–	–	–	258[i9]	–	[718]
	–	–	–	350–440	–	–	[452]
	462[b]	102[b]	154[b]	–	224[b]	0.20[b]	[19]
	–	–	–	387[j1]	223[j1]	–	[719]
	–	–	–	–	245[g2]	–	[720]
	–	–	–	–	220[g3]	–	[720]
	458[e8,j2]	101[e8,j2]	156[e8,j2]	–	218[e8,j2]	–	[20]
	–	–	–	–	281[b]	–	[21]
	–	–	–	–	220[b]	–	[282]
	–	–	–	407[j3]	223[j3]	0.20[j3]	[721]
	452[j4]	92[j4]	112[j4]	341[j4,j5]	212[j4,j5]	0.22[j4]	[52, 722]
	455[h2]	101[h2]	150[h2]	387[h2,j6]	219[h2,j6]	–	[724]
	457[b]	98[b]	157[b]	395[b,j7]	224[b,j7]	–	[22]
	468[j8]	100[j8]	159[j8]	433[j8,j9]	222[j8,j9]	0.18[j8]	[725]
	483[k1]	109[k1]	142[k1]	443[k1,k2]	234[k1,k2]	0.18[k1]	[725]
	468[k3]	102[k3]	148[k3]	390[k3,c3]	224[k3,c3]	0.21[k3]	[24]
	–	–	–	400[k4]	–	0.20	[54]
	–	–	–	415[k5,k6]	234[k5,k6]	0.21[k5]	[726]
	557[k7]	95[k7]	155[k7]	–	248[k7]	–	[727]

(continued)

Table 5.17 (continued)

Composition	Stiffness coefficients c_{ij}			Young's modulus E, GPa	Bulk modulus K, GPa	Poisson's ratio ν	Reference
	c_{11}, GPa	c_{12}, GPa	c_{44}, GPa				
	446[b]	103[b]	153[b]	–	217[b]	–	[728]
	–	–	–	240 ± 10[k8]	–	–	[729]
	–	–	–	290 ± 13[k9]	–	–	[729]
	–	–	–	320 ± 17[l1]	–	–	[729]
	–	–	–	–	220[l2]	–	[730]
	458[b]	104[b]	152[b]	420[b,l3]	222[b,l3]	0.19[b]	[731]
	488[j2,l4]	108[j2,l4]	160[j2,l4]	413[j2,l4,l5]	235[j2,l4,l5]	0.21[j2,l4]	[283, 732]
	375[l4,l6]	204[l4,l6]	253[l4,l6]	406[l4,l6,l7]	261[l4,l6,l7]	0.24[l4,l6]	[283, 732]
	–	–	–	–	222[l8]	–	[284]
	482[l8]	113[l8]	146[l8]	392[j6,l8]	236[j6,l8]	0.22[l8]	[285]
	503[l9]	99[l9]	171[l9]	–	234[l9]	–	[733]
	438[b]	93[b]	161[b]	–	208[b]	–	[733]
	446[b]	104[b]	138[b]	407[b,m1]	218[b,m1]	0.19[b]	[376]
	452[m2]	105[m2]	150[m2]	385[m2,m3]	221[m2,m3]	0.21[m2]	[734]
	–	–	–	377[m4,m5]	216[m4,m5]	–	[734]
	487[m6]	107[m6]	152[m6]	403[m6,g5]	234[m6,g5]	0.21[m6]	[734]
	–	–	–	395[f8,m7]	229[f8,m7]	–	[734]
	511[m8]	111[m8]	156[m8]	418[m8,m9]	244[m8,m9]	0.22[m8]	[734]
	–	–	–	399[l7,n1]	233[l7,n1]	–	[734]
	490[b]	112[b]	147[b]	–	238[b,c7]	–	[735]
	–	–	–	310[n2]	–	–	[617]
	–	–	–	380–407[n3]	–	–	[666]
	–	–	–	–	231[l8]	–	[808]
	–	–	–	405[n4]	–	–	[1382]
	–	–	–	–	233–253[n5]	–	[1390]
	–	–	–	398[n6,n7,n8]	–	0.19	[1479]
	513[l9]	113[l9]	163[l9]	428[l9,n9]	246[l9,n9]	0.21[l9]	[1516]
	452[b]	107[b]	155[b]	391[b,o1]	222[b,o1]	0.21[b]	[1516]
	433[g8]	104[g8]	155[g8]	382[g8,o2]	214[g8,o2]	0.20[g8]	[1516]
	475[g7]	134[g7]	156[g7]	398[g7,o3]	247[g7,o3]	0.23[g7]	[1516]
	–	–	–	–	210[b,o2]	–	[1517]
	–	–	–	549–550[d1]	–	0.10	[33, 365, 674, 692]

[a]For isotropic (or quasi-isotropic) materials [3]: $E = 2G(1 + \nu)$, $E = 3K(1 - 2\nu)$, $K = c_{12} + 2c_{44}/3$, $G = c_{44}$; the condition for the isotropy is given in Eq. (I-2.18)
[b]Calculated on the basis of density-functional theory (DFT) with generalized gradient approximation (GGA) using the Perdew-Burke-Ernzerhof (PBE) functional
[c]Calculated value of G = 108 GPa
[d]Sintered, corrected to the poreless state (true porosity – 5–10%)
[e]Measured experimentally G = 142 GPa
[f]Calculated value of G = 120 GPa
[g]Sintered, porosity – 12%, contents: non-combined C < 0.10%, O – 1.20%, N – 0.22%, W – 0.35%; sonic velocity measurement method
[h]Hot-pressed, porosity <4%, content non-combined C – 1.6–2.5%
[i]Sintered (in vacuum) and annealed, porosity – 21%, mean grain size – 18 μm, contents: non-combined C – 0.01%, O – 0.26%, N – 0.20%

(continued)

Table 5.17 (continued)

[j]Measured experimentally G = 125 GPa
[k]Sintered, porosity – 8–10%
[l]G = 132 GPa
[m]Calculated value of G = 137 GPa
[n]Sintered, porosity – 10%, contents: non-combined C < 0.10%, O – 1.60%, N – 0.035%, W – 0.35%; sonic velocity measurement method
[o]Theoretical estimation; for $ZrC_{0.80}$ the same values determined by Zr–Zr and Zr–C bonds are 74 GPa and 610 GPa, respectively, and for $ZrC_{0.96–0.98}$ the same values determined by Zr–Zr and Zr–C bonds are 110 GPa and 623 GPa, respectively
[p]Calculated on the basis of DFT by the means of the exact muffin-tin orbitals within coherent potential approximation (EMTO-CPA) method using GGA for the total energies evaluation
[q]Calculated value of G = 144 GPa
[r]Sintered, porosity – 4–6%
[s]G = 148 GPa
[t]Measured experimentally G = 146 GPa
[u]Sintered, porosity – 7%, contents: non-combined C < 0.10%, O – 1.00%, N – 0.03%, W – 0.20%; sonic velocity measurement method
[v]Sintered (in vacuum) and annealed, porosity – 15%, mean grain size – 15 μm, contents: non-combined C – 0.01%, O – 0.62%, N – 0.01%
[w]Measured experimentally G = 151 GPa
[x]Zone refined materials
[y]Calculated value of G = 147 GPa
[z]Calculated value of G = 147 GPa
[a1]Single crystal materials; sonic resonance method
[a2]Sintered in vacuum, contents: non-combined C – 0.15–0.20%, O – 0.10–0.18%
[a3]G = 155 GPa
[a4]Measured experimentally G = 150 GPa
[a5]Sintered, extrapolated to zero porosity
[a6]Sintered (in vacuum) and annealed, porosity – 14%, mean grain size – 11 μm, contents: non-combined C – 0.10%, O – 0.50%, N – 0.01%
[a7]Measured experimentally G = 158 GPa
[a8]Sintered, porosity – 10%
[a9]Sintered, porosity – 4–5%, content non-combined C < 0.1%
[b1]Sintered, porosity 6–8%, contents: non-combined C – 0.5%, W – 0.2%
[b2]Calculated values: G = 172 GPa (isotropic), E_{001} = 438 GPa, E_{011} = 398 GPa, E_{111} = 386 GPa, G_{001} = 159 GPa, G_{011} = 172 GPa, G_{111} = 177 GPa, $2c_{44}/(c_{11} - c_{12})$ = 0.85
[b3]Measured experimentally G = 154 GPa (according to Frost and Ashby [659]: G = 173 GPa)
[b4]Hot-pressed, corrected to the poreless state (true porosity – 8%); calculated from the data on Hertizan indentation measurements
[b5]G = 167 GPa (according to Frost and Ashby [659]: G = 173 GPa)
[b6]Hot-pressed, contents: non-combined C – 0.40%, O – 0.07–0.09%, N – 0.10%, Fe – 0.1%, corrected to the poreless state (true porosity – 3%); sonic resonance method
[b7]Measured experimentally G = 162 GPa
[b8]Sintered (in vacuum) and annealed, porosity – 6%, mean grain size – 10 μm, contents: non-combined C – 0.55%, O – 0.13%, N – 0.07%
[b9]Measured experimentally G = 160 GPa
[c1]Sintered, porosity – 6%, contents: non-combined C – 0.13%, O – 0.08%, N – 0.01%, W – 0.01%; sonic velocity measurement method
[c2]Sintered, porosity – 6%, contents: O $\leq$ 0.05%, N – 0.05%, W – 0.2–0.3%
[c3]Calculated value of G = 161 GPa
[c4]G = 185 GPa
[c5]Theoretical estimation; the same values for Zr–Zr and Zr–C bonds are 110 GPa and 623 GPa, respectively

(continued)

Table 5.17 (continued)

[c6]Hot-pressed, corrected to the poreless state (true porosity – up to 12%), mean grain size – 5–20 μm; ultrasonic wave velocity measurements
[c7]G = 163 GPa
[c8]Single crystal materials, Bernstein's data
[c9]Calculated by Brown and Kempter using Kröner's averaging technique for Voigt-Reuss calculations, calculated value of G = 163 GPa
[d1]Corrected to the poreless state
[d2]Hot-pressed, porosity – 3%
[d3]Measured experimentally G = 170 GPa
[d4]Extrapolated to stoichiometric pure carbide phase from the experimental data on compositions containing non-combined carbon
[d5]Estimated by means of density-functional perturbation theory (DFPT) calculations
[d6]G = 161 GPa
[d7]Estimation from phonon dispersion curves
[d8]Hot-pressed, porosity – 6%; calculated from depth-sensing indentation (DSI) measurements
[d9]Hot-pressed, porosity – 8%; determined by acoustic method
[e1]Sintered, porosity – 10%, mean grain size – 10 μm
[e2]Calculated from thermal experimental measurements of elastic constants
[e3]Calculated from high-pressure experiments of Champion and Drickamer [723]
[e4]On the basis of interconsistency of the physical properties of transition metal monocarbides
[e5]Porosity – 8%
[e6]G = 124 GPa
[e7]Sintered, porosity ~5%
[e8]Calculated on the basis of DFT using full-potential linearized augmented-plane-wave (FP-LAPW) method with GGA
[e9]Calculated value of G = 168 GPa
[f1]Zero temperature, zero pressure
[f2]Spark-plasma sintered, porosity – 16% (for the same post-annealed material E = 236 GPa)
[f3]At 0 K (at 2000 K: K = 193 GPa)
[f4]Assuming uniform strain throughout the polycrystalline single-phase materials
[f5]Calculated value of G = 223 GPa
[f6]Assuming uniform stress throughout the polycrystalline single-phase materials
[f7]Calculated value of G = 204 GPa
[f8]Calculated value of G = 163 GPa
[f9]Calculated value of G = 132 GPa
[g1]Extrapolated to the stoichiometric composition from the experimental data obtained by Chang and Graham [49]
[g2]Calculated on the basis of DFT using local density approximation (LDA)
[g3]Calculated on the basis of DFT using using plane-wave pseudopotential (PWPP) method GGA
[g4]Calculated on the basis of DFT using empirical (force-based many-body interatomic) potential
[g5]Calculated values of G = 166 GPa
[g6]Magnetron-sputtered nanocrystalline films
[g7]Calculated on the basis of DFT with GGA using Perdew's functional
[g8]Calculated on the basis of DFT with GGA using the revised version of the Perdew-Burke-Ernzerhof (RPBE) functional
[g9]Calculated on the basis of DFT with Wu-Cohen GGA exchange-energy functional
[h1]Calculated on the basis of DFT using FP-LAPW method with LDA
[h2]Calculated using the Debye–Grüneisen model combined with *ab initio* calculations
[h3]Calculated value of G = 124 GPa
[h4]Summarized values on the basis of critical review of available experimental data
[h5]G = 165 GPa

(continued)

Table 5.17 (continued)

[h6]Spark plasma sintered, porosity – 1–2%, mean grain size – 5–13 μm; from nanoindentation measurements
[h7]Calculated on the basis of interionic potential theory with modified ionic charge
[h8]Calculated value of G = 144 GPa
[h9]Calculated value of isotropic E = 435 GPa
[i1]Calculated value of G = 186 GPa
[i2]For isotropic structure calculated value of ν = 0.18
[i3]Calculated value of G = 185 GPa
[i4]Calculated value of G = 185 GPa
[i5]Calculated value of G = 184 GPa
[i6]Calculated on the basis of linear combination of atomic orbitals (LCAO) method using GGA and Perdew-Wang potential
[i7]Calculated value of G = 170 GPa
[i8]Calculated using the quasi-harmonic Debye model combined with *ab initio* calculations
[i9]Calculated on the basis of DFT with GGA using the plane-wave pseudopotential code Dacapo and Perdew-Wang functional
[j1]G = 160 GPa
[j2]At zero temperature and pressure conditions
[j3]G = 170 GPa
[j4]Calculated on the basis of a modified interaction potential model with covalency ($MIPMC_v$)
[j5]Calculated value of G = 140 GPa
[j6]Calculated value of G = 160 GPa
[j7]Calculated value of G = 165 GPa
[j8]Calculated on the basis of three body force potential (TBFP) model
[j9]Calculated values of G = 169 GPa and κ = 4.5 TPa^{-1}
[k1]From *ab initio* calculations
[k2]Calculated values of G = 159 GPa and κ = 4.3 TPa^{-1}
[k3]Calculated on the basis of DFT using full-potential linearized augmented plane-wave plus local orbital (FP-LAPW+LO) method with GGA and PBE funtional
[k4]κ = 4.50 TPa^{-1}
[k5]Spark plasma sintered, porosity – 1%
[k6]G = 172 GPa
[k7]Calculated on the basis of DFT using FP-LAPW+LO method with the Perdew-Wang LDA
[k8]Pulsed laser deposited thin films, mean grain size – 6 nm, microstrain – 2.0%, content O < 0.5%; calculated from nanoindentation measurements
[k9]Pulsed laser deposited thin films, mean grain size – 12 nm, microstrain – 1.7%, content O < 0.5%; calculated from nanoindentation measurements
[l1]Pulsed laser deposited thin films, mean grain size – 19 nm, microstrain – 1.6%, content O < 0.5%; calculated from nanoindentation measurements
[l2]By means of extended Hückel tight-binding band electronic structure calculations
[l3]Calculated value of G = 162 GPa
[l4]Calculated by formulating effective interatomic interaction potential
[l5]Calculated value of G = 171 GPa
[l6]At ~300 K
[l7]Calculated value of G = 164 GPa
[l8]Calculated on the basis of DFT by PAW method with GGA using PBE functional
[l9]Calculated on the basis of DFT with LDA using Ceperley-Alder-Perdew-Zunger (CAPZ) functional
[m1]Calculated value of G = 150 GPa
[m2]Calculated on the basis of DFT with GGA using the PBE parametrization from the stress-strain relationship
[m3]Calculated value of G = 159 GPa
[m4]Calculated on the basis of DFT with GGA using the PBE parametrization from the energy-volume fitting

(continued)

Table 5.17 (continued)

[m5]Calculated values of G = 156 GPa
[m6]Calculated on the basis of DFT with GGA-PBE exchange correlations for solids from the stress-strain relationship
[m7]Calculated on the basis of DFT with GGA-PBE exchange correlations for solids from the energy-volume fitting
[m8]Calculated on the basis of DFT using LDA from the stress-strain relationship
[m9]Calculated value of G = 172 GPa
[n1]Calculated on the basis of DFT using LDA from the energy-volume fitting
[n2]Sintered, porosity – 7%, contents: non-combined C (graphite) – 0.9%, O $\leq$ 0.05%, N – 0.05%, W – 0.2–0.3%
[n3]Sintered, porosity – 4–6%, contents: non-combined C – 0.5%, O – 0.05–0.07%, N – 0.02–0.07%, W < 0.05%, Fe < 0.01%
[n4]Experimentally measured by the method of continuous impression of an indenter
[n5]Calculated by all-electron pseudopotentials with pseudo-wave functions
[n6]Hot-pressed, porosity – 2%, content O – 0.36%; measured by an impulse excitation technique
[n7]At 1300 °C: E = 313 GPa
[n8]Measured experimentally G = 167 GPa
[n9]Calculated value of G = 177 GPa
[o1]Calculated value of G = 162 GPa
[o2]Calculated value of G = 159 GPa
[o3]Calculated value of G = 161 GPa

The variations of Young's E, Coulomb's (shear) G and bulk (compression) K moduli and Poisson's ratio v with temperature for the ZrC_{1-x} phases with different deviations from the stoichiometry are shown in Fig. 5.32 and the variations with carbon content – in Fig. 5.33; the elastic properties (stiffness coefficients c_{ij} and moduli E, K and Poisson's ratio v) of various ZrC_{1-x} materials are listed in Table 5.17. For $ZrC_{\sim 1.0}$ the approximate average rate of decay of Young's modulus with temperature $1/E_{298K} \times dE/dT = -1.2 \times 10^{-4}\ K^{-1}$ [52], that is a bit different to the data presented earlier [33, 365, 674, 692] by the following equation (corrected to porosity)

$$E = 550\,[1 - (7.0 \times 10^{-5})t], \tag{5.85}$$

where t is temperature, °C. However, as it was reported by Baranov et al. [670], the temperature dependence of Young's modulus E of polycrystalline ZrC_{1-x} materials with various deviations from stoichiometry is more complicated: below $\sim$1700 °C it is fairly well represented by Wachtman's equation:

$$E = E_0 - BT \exp(-T_0/T), \tag{5.86}$$

where E_0 is the value of E at T = 0 K, B and T_0 are constants affected by the composition of materials (T_0 decreases as the composition deviates further from the stoichiometry) and T is temperature, K, and at the temperatures above $\sim$1700 °C there is the deviation of dependence graph from the logarithmic linearity, which can be representing by adding to the above equation an exponential term [670]:

$$E = E_0 - BT\ \exp(-T_0/T) - C\ \exp(-H_a/RT), \tag{5.87}$$

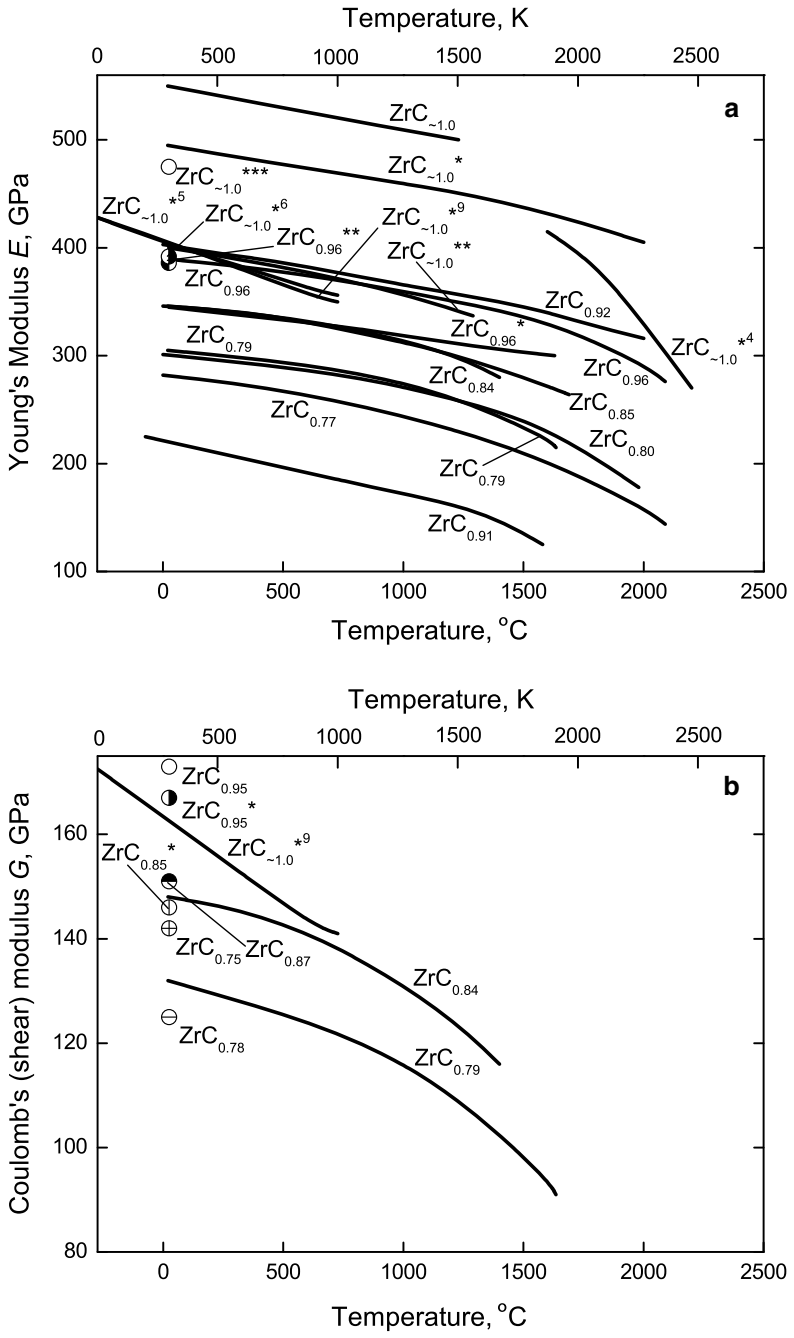

Fig. 5.32 (continued)

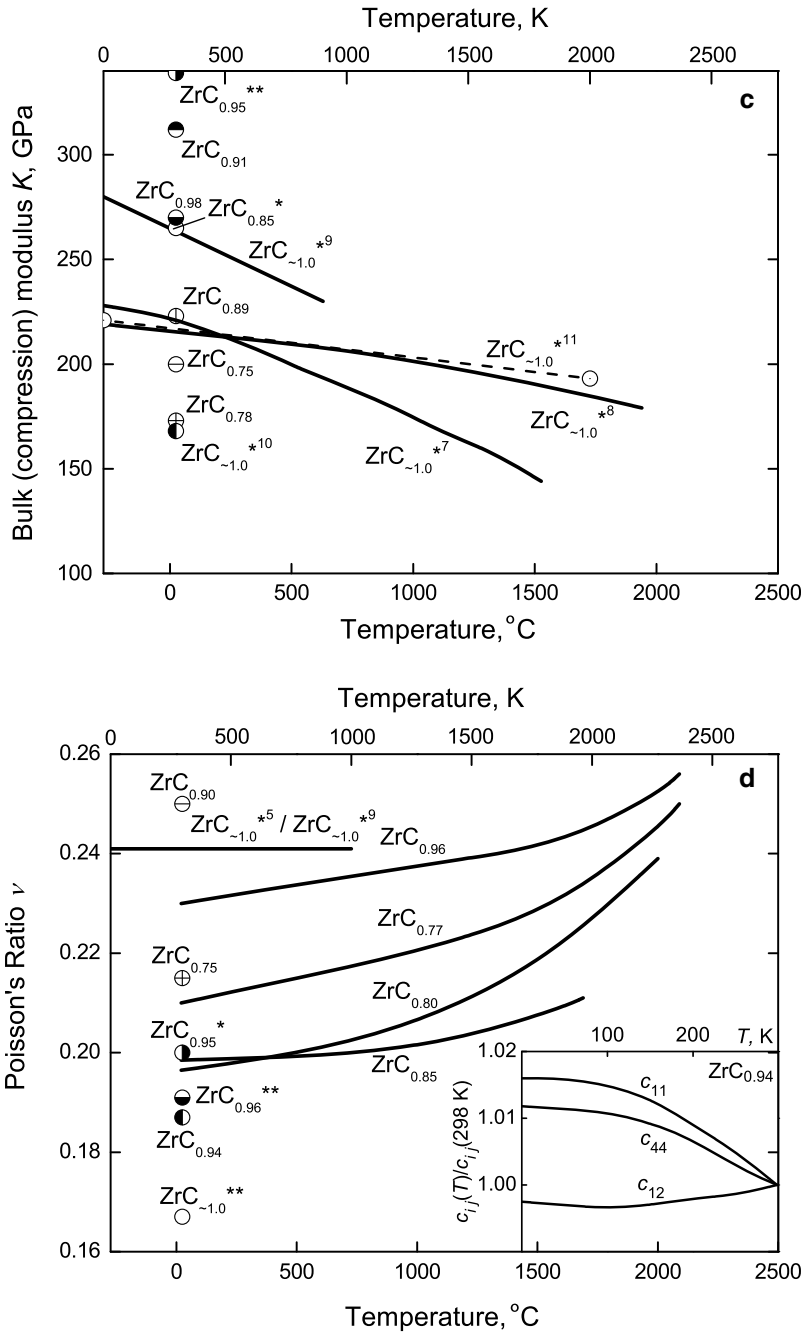

Fig. 5.32 (see caption on page 510)

◂**Fig. 5.32** Variations of Young's modulus E (**a**), Coulomb's (shear) modulus G (**b**), bulk (compression) modulus K (**c**) and Poisson's ratio ν (**d**) of zirconium carbide ZrC_{1-x} materials with temperature: sintered $ZrC_{0.75}$, $ZrC_{0.85}$* and $ZrC_{0.95}$** (porosity – 5–10%), corrected to the poreless state [125, 675, 676]; sintered $ZrC_{0.77}$ (porosity – 12%, contents: non-combined C < 0.10%, O – 1.20%, N – 0.22%, W – 0.35%), $ZrC_{0.80}$ (porosity – 10%, contents: non-combined C < 0.10%, O – 1.60%, N – 0.035%, W – 0.35%), $ZrC_{0.85}$ (porosity – 7%, contents: non-combined C < 0.10%, O – 1.00%, N – 0.03%, W – 0.20%) and $ZrC_{0.96}$ (porosity – 6%, contents: non-combined C – 0.13%, O – 0.08%, N – 0.01%, W – 0.01%) [123, 670, 689]; sintered (in vacuum) and annealed $ZrC_{0.78}$ (porosity – 21%, mean grain size – 18 μm, contents: non-combined C – 0.01%, O – 0.26%, N – 0.20%) [4, 132]; sintered $ZrC_{0.79}$ (porosity – 8–10%) and $ZrC_{0.84}$ (porosity – 4–6%) [685]; sintered (in vacuum) and annealed $ZrC_{0.87}$ (porosity – 15%, mean grain size – 15 μm, contents: non-combined C – 0.01%, O – 0.62%, N – 0.01%) [4, 132]; single crystal $ZrC_{0.89}$ [49, 705, 706]; sintered (in vacuum) $ZrC_{0.90}$ (contents: non-combined C – 0.15–0.20%, O – 0.10–0.18%) [277]; sintered (in vacuum) $ZrC_{0.91}$ (porosity – 18%) [125, 675, 676]; sintered $ZrC_{0.92}$ (porosity – 10%) [575]; single crystal $ZrC_{0.94}$ [49]; $ZrC_{0.95}$ (mean grain size – 10 μm) [659]; hot-pressed $ZrC_{0.95}$*, (porosity – 8%), calculated from the data on Hertizan indentation measurements, corrected to porosity [96, 662]; sintered $ZrC_{0.96}$* (porosity – 7%, content O – 0.04%) [4]; hot-pressed $ZrC_{0.96}$** (porosity – 3%, contents: non-combined C – 0.40%, O – 0.07–0.09%, N – 0.10%, Fe – 0.1%), corrected to the poreless state [82]; zone refined $ZrC_{0.98}$ [277]; sintered $ZrC_{\sim 1.0}$, corrected to the poreless state [33, 365, 674]; $ZrC_{\sim 1.0}$*, corrected to porosity [3]; hot-pressed $ZrC_{\sim 1.0}$** (porosity – 4%) [308]; $ZrC_{\sim 1.0}$***, extrapolated to stoichiometric pure carbide phase from the experimental data on compositions containing non-combined carbon [673]; die-extruded and sintered $ZrC_{\sim 1.0}$*[4], calculated from stress relaxation tests [625–627]; $ZrC_{\sim 1.0}$*[5], *ab initio* calculated [283]; $ZrC_{\sim 1.0}$*[6], extrapolated to zero porosity [680]; $ZrC_{\sim 1.0}$*[7], theoretically calculated on the basis of linear combination of atomic orbitals (LCAO) method [69]; $ZrC_{\sim 1.0}$*[8], theoretically calculated using the Debye–Grüneisen model combined with *ab initio* calculations [724]; $ZrC_{\sim 1.0}$*[9], *ab initio* calculated [283]; $ZrC_{\sim 1.0}$*[10], calculated [49, 82, 677, 678]; $ZrC_{\sim 1.0}$*[11], calculated on the basis of DFT by the means of the exact muffin-tin orbitals within coherent potential approximation (EMTO-CPA) method [699] (*Inset in* **d** – variations of normalized elastic constants of single crystal $ZrC_{0.94}$ with low- and ultra-low temperatures in $c_{ij}(T)/c_{ij}(298\ \mathrm{K})$ – T, K scale [49])

where T is temperature, K, C is the constant, which varies from 3.9×10^3 GPa for $ZrC_{0.77}$ to 4.9×10^5 GPa for $ZrC_{0.96}$, R is the gas constant and H_a is a quantity with the significance of the effective activation energy of the thermally activated processes, which gives a more than linear fall of the modulus and varies from 85.4 kJ mol^{-1} for $ZrC_{0.77}$ to 198 kJ mol^{-1} for $ZrC_{0.96}$. Ajami and MacCrone [688] proposed Wachtman's equation for bulk (compression) K modulus of quasi-stoichiometric zirconium monocarbide $ZrC_{\sim 1.0}$ in the wide temperature range:

$$K = 256 - (1.94 \times 10^{-2})T \exp(-257/T), \tag{5.88}$$

where T is temperature, K. According to the calculations on the basis of DFT with LDA produced by Krasnenko and Brik [733] bulk (compression) modulus K of $ZrC_{\sim 1.0}$ arises at elevated hydrostatic pressures p (up to 50 GPa) linearly

$$K = 238.0 + 3.65p. \tag{5.89}$$

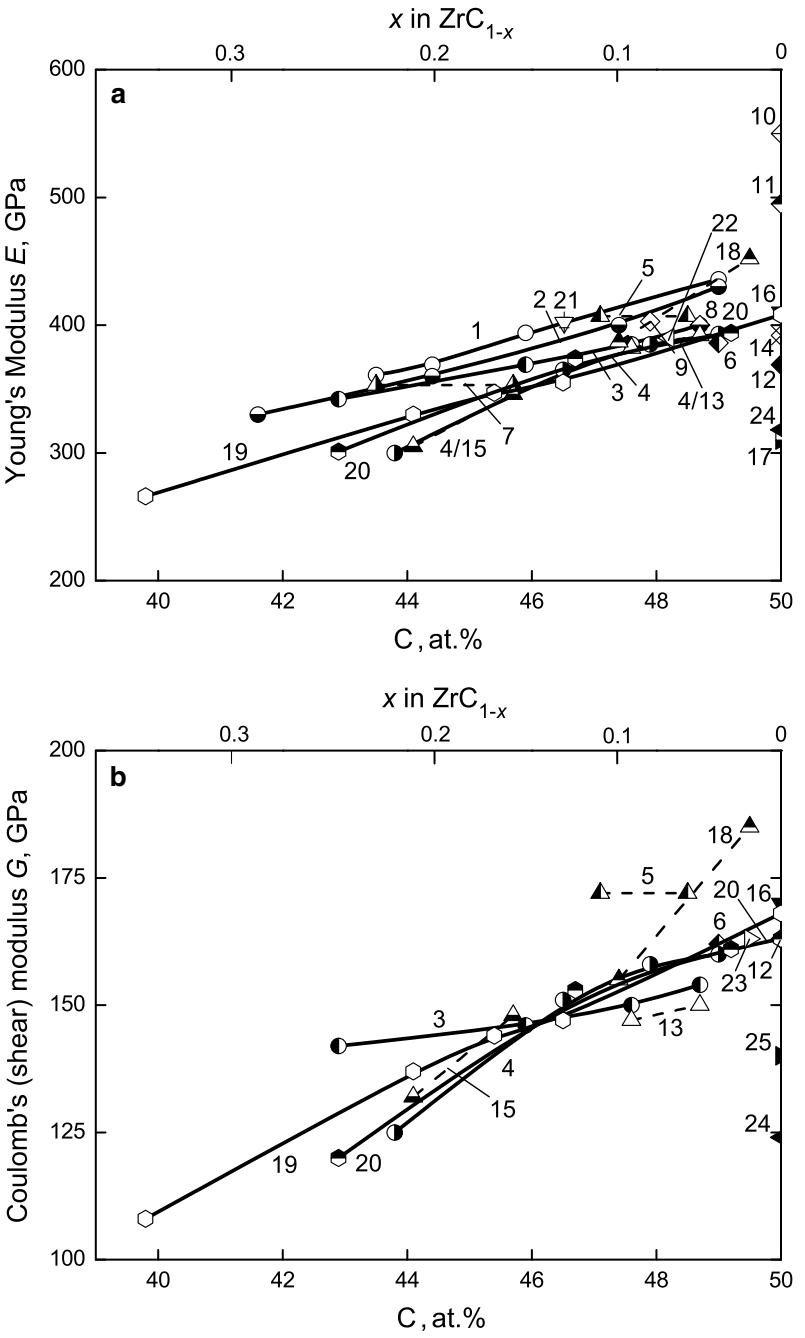

Fig. 5.33 (continued)

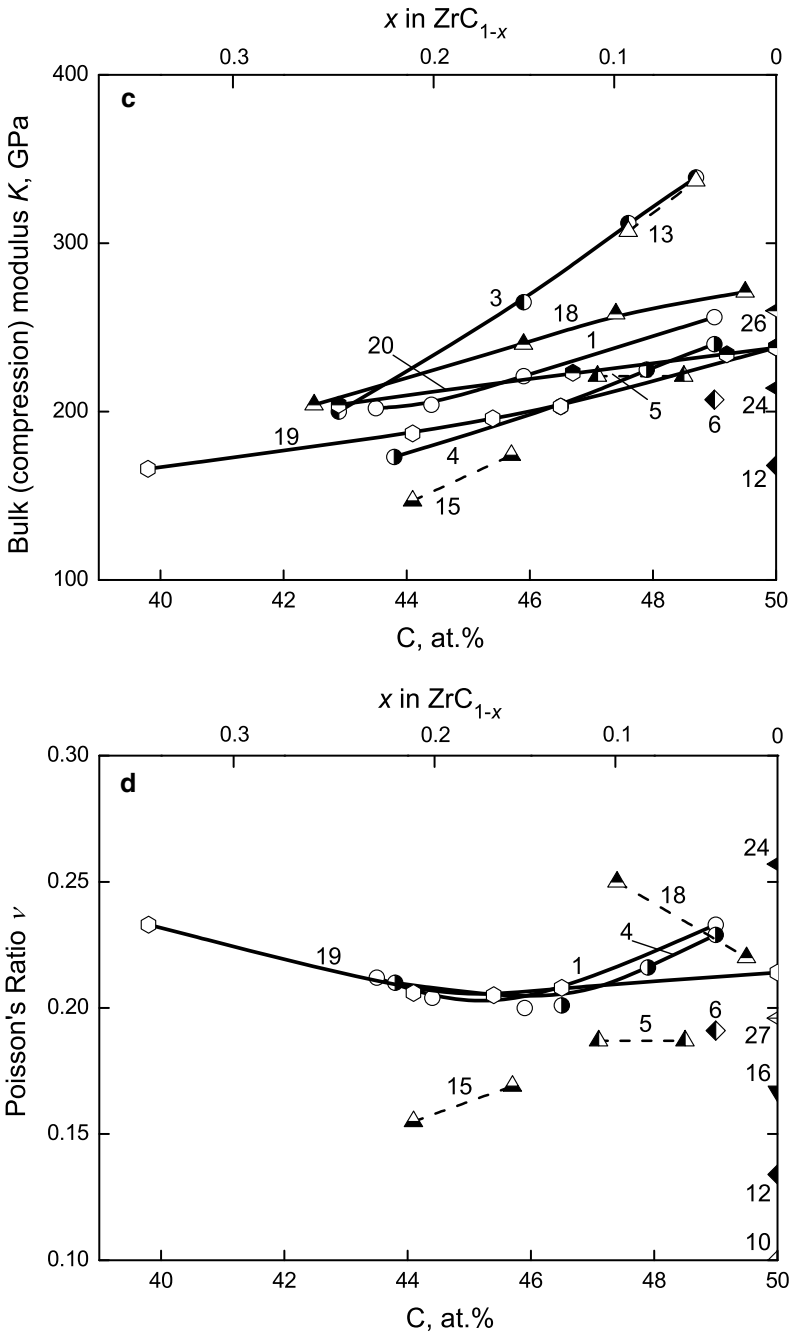

Fig. 5.33 (see caption on page 513)

◂**Fig. 5.33** Variations of Young's modulus *E* (**a**), Coulomb's (shear) modulus *G* (**b**), bulk (compression) modulus *K* (**c**) and Poisson's ratio *v* (**d**) of zirconium carbide ZrC_{1-x} with deviation from the stoichiometry within the homogeneity range at ambient temperatures: 1 – sintered materials, corrected to porosity [123, 670, 689]; 2 – sintered materials [1]; 3 – sintered materials, corrected to the poreless state (true porosity – 5–10%) [125, 675, 676]; 4 – sintered (in vacuum) and annealed materials, porosity – 6–21%, mean grain size – 10–18 μm, contents: non-combined C – 0.10–0.55%, O – 0.13–0.79%, N – 0.01–0.03% [4, 132]; 5 – single crystal materials, calculated on the basis of elastic constants [11, 49]; 6 – hot-pressed materials, corrected to the poreless state (true porosity – 3%) [11, 82]; 7 – hot-pressed materials, porosity <4%, contents: non-combined C – 1.6–2.5% [356, 634]; 8 – sintered materials, porosity – 8% [662]; 9 – sintered materials, porosity – 10% [575]; 10 – sintered materials, corrected to porosity [33, 365, 674]; 11 – corrected to porosity [3]; 12 – single crystal materials [49, 82, 677, 678]; 13 – sintered materials, extrapolated to zero porosity [679]; 14 – extrapolated to zero porosity [680]; 15 – sintered materials, porosity – 4–10% [685]; 16 – hot-pressed materials, porosity – 3% [308]; 17 – hot-pressed materials, porosity – 8% [41]; 18 – sintered (in vacuum) materials, contents: non-combined C – 0.15–0.20%, O – 0.10–0.18% [277]; 19 – calculated on the basis of density-functional theory (DFT) [696]; 20 – calculated on the basis of DFT [735]; 21 – zone refined materials [96, 681]; 22 – sintered materials, porosity – 4–5% [624]; 23 – hot-pressed, corrected to the poreless state (true porosity – up to 12%), mean grain size – 5–20 μm [703]; 24 – [39]; 25 – calculated on the basis of a modified interaction potential model with covalency ($MIPMC_v$) [722]; 26 – calculated on the basis of DFT [715]; 27 – [721]

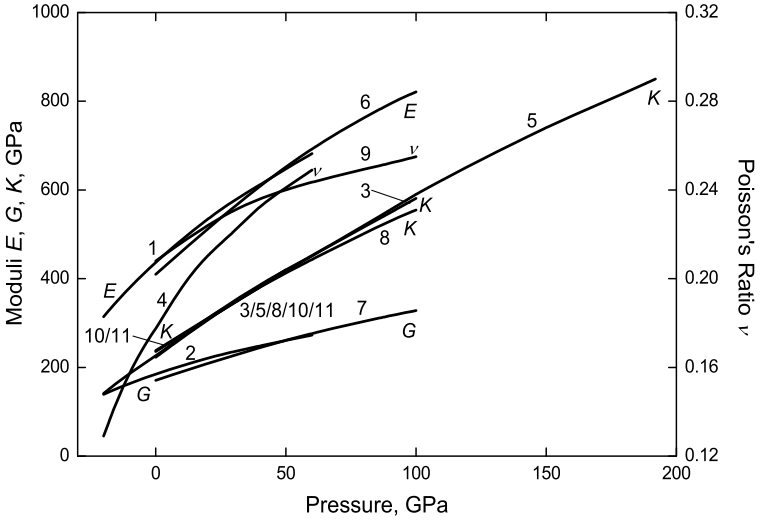

Fig. 5.34 Variations of elastic properties (1, 6 – Young's modulus *E*; 2, 7 – Coulomb's (shear) modulus *G*; 3, 5, 8, 10–11 – bulk (compression) modulus *K* and 4, 9 – Poisson's ratio *v*) of quasi-stoichiometric zirconium monocarbide $ZrC_{\sim 1.0}$ with higher pressure based on *ab initio* and other theoretical calculations: 1–4 – by Fu et al. [69], 5 – by Zhu et al. [281], 6–9 – by Varshney et al. [283, 732], 10 – by Gautam and Kumar [285], 11 – by Krasnenko and Brik [733]

Table 5.18 Thermal shock/stress resistance (thermal strength) testing data on single phase zirconium monocarbide ZrC_{1-x} materials

Materials and its characterization	Specimen shape and dimensions, mm	Method of thermal loading (testing)	Experimental (or calculated) results[a]	Reference
$ZrC_{0.83–0.85}$, hot-pressed, porosity – 2%, content non-combined C – 4.1–4.3%	Disks: diameter – 50.8, thickness – 6.35	Quenching in air stream (~76 m s^{-1}) in parallel direction to the flat surfaces, full cycle time – 15 min	No disintegration after 22 heating-cooling cycles from 980 to 30 °C (excessive oxidation was noted)	[349, 350]
$ZrC_{0.93}$, sintered, porosity – 10%, contents: non-combined C – 0.2%, N – 0.04%, W – 0.3%	Ring-shaped disks: outer diameter – 50, inner diameter – 44, height – 10	Steady-state heating along the outer surface	R = 40–50 K (calculated), R' = 2.3 ± 0.4 kJ m^{-1} s^{-1} (experimental)	[3, 4, 740]
$ZrC_{0.93}$, sintered	Rods: diameter – 3, length – 30	Quenching in water	R = 60 K	[334, 335, 629]
$ZrC_{0.93}$, sintered	Disks: diameter – 25, height – 3	Quenching in water	R = 50 K	[334, 335, 629]
$ZrC_{0.93}$, sintered	Disks: diameter – 35, height – 2.5	Unsteady-state heating in molten tin	R = 61 K	[334, 335, 617, 629]
$ZrC_{0.93}$, sintered	Disks: diameter – 20, height – 1	Surface electron-beam heating	R = 65 K	[334, 335, 629, 687]
$ZrC_{0.93–0.95}$, sintered, porosity – 4–5%, content non-combined C < 0.1%	Disks: diameter – 24–26, thickness – 1–2.4; rods: diameter – 2.2, length – 50–80	Surface electron-beam heating with the usage of special diaphragms	At 50 °C: R = 65 K, R' = 1.3 kJ m^{-1} s^{-1}; in cryogenic conditions: at −170 °C: R = 106 K, R' = 2.6 kJ m^{-1} s^{-1}	[334, 335, 624, 687, 741]
$ZrC_{0.93–0.96}$, sintered, porosity – 4–80%	–	–	Extrapolated to zero porosity: R = 45 K, R' = 1.9 kJ m^{-1} s^{-1}; (for samples with 80% porosity: R = 70 K, R' = 0.08 kJ m^{-1} s^{-1})	[330, 334, 335]
$ZrC_{0.94}$, sintered	Cylinders: diameter – 15, height – 45	Induction heating	R = 50 K	[334, 335, 629]

(continued)

Table 5.18 (continued)

Materials and its characterization	Specimen shape and dimensions, mm	Method of thermal loading (testing)	Experimental (or calculated) results[a]	Reference
$ZrC_{0.95}$, sintered, porosity – 5%, non-combined C < 0.1%	Cylinders: diameter – 55, height – 50	Unsteady-state radiative heating	Variation of σ_{φ}^{b} from 34 to 81 MPa	[3, 4]
$ZrC_{0.95}$, sintered, porosity – 10%	Continuous and ring-shaped disks: outer diameter – 16–52, ratio of inner diameter to outer diameter – 0–0.9	Unsteady-state heating in molten tin, cooling in water	R = 40–70 K (complete disintegration during heating), R = 40 K (appearance of surface cracks during cooling)	[3, 4, 739, 744]
$ZrC_{0.95–0.97}$, sintered, porosity – 4–6%, contents: O – 0.05–0.07%, N – 0.02–0.07%, W < 0.05%, Fe < 0.01%	Ring-shaped disks: inner diameter – 4–6, ratio of inner diameter to outer diameter – 0.16	Unsteady-state heating in molten tin	$R' = 1.9$ kJ m^{-1} s^{-1}	[666]
$ZrC_{0.96}$, sintered, porosity – 8%	Cylinders: diameter – 20, height – 50	Unsteady-state radiative heating	R = 30 K (R_l = 40 K[c], σ_z = 96 MPa[d])	[3, 4]
$ZrC_{0.96}$, sintered, porosity – 6%, contents: O ≤ 0.05%, N – 0.05%, W – 0.2–0.3%	Disks with a central hole: inner diameter – 5	Unsteady-state heating along the lateral surface of central hole in molten tin	R = 49 K, $R' = 1.4$ kJ m^{-1} s^{-1}, $R''' = 4.2$ MPa^{-1}, $R'''' = 0.05$ mm	[334, 335, 617]
$ZrC_{0.96}$, sintered	Cylinders: diameter – 15, height – 45	Unsteady-state radiative heating	R = 57 K	[334, 335, 629, 743]
$ZrC_{\sim 1.0}$, hot-pressed, porosity – 9%	Spheres: diameter – 50.8	Unsteady-state (transient) state radiative heating	Observed experimentally temperature required to initiate fracture – 1450 °C	[308]
$ZrC_{\sim 1.0}$, gas-phase deposited, contents: O + N ≤ 0.3%	Hollow cylinders: wall thickness – 0.8–1.2, height – 20	Direct resitance heating with 400 K s^{-1}, cooling by inert gas blow, cycle time – 3 min	No disintegration after 16 heating-cooling cycles from 2000 to 50 °C, disintegration during the first heating-cooling cycle from 2400 to 50 °C	[4, 539]
$ZrC_{\sim 1.0}$, gas-phase deposited, contents: non-combined C – 1.0%, O + N ≤ 0.3%	Hollow cylinders: wall thickness – 0.8–1.2, height – 20	Direct resitance heating with 400 K s^{-1}, cooling by inert gas blow, cycle time – 3 min	Disintegration during 4 heating-cooling cycles from 2400 to 50 °C	[4, 539]

(continued)

Table 5.18 (continued)

Materials and its characterization	Specimen shape and dimensions, mm	Method of thermal loading (testing)	Experimental (or calculated) results[a]	Reference
$ZrC_{\sim 1.0}$, sintered, porosity – 7%, contents: non-combined C (graphite) – 0.9%, O $\leq$ 0.05%, N – 0.05%, W – 0.2–0.3%	Disks with a central hole: inner diameter – 5	Unsteady-state heating along the lateral surface of central hole in molten tin	R = 25 K, R' = 0.6 kJ m^{-1} s^{-1}, R''' = 15.5 MPa^{-1}, R'''' = 0.17 mm	[617]
$ZrC_{\sim 1.0}$, sintered, porosity – 4–6%, contents: non-combined C – 0.5%, O – 0.05–0.07%, N – 0.02–0.07%, W < 0.05%, Fe < 0.01%	Ring-shaped disks: inner diameter – 4–6, ratio of inner diameter to outer diameter – 0.16	Unsteady-state heating in molten tin	R' = 3.2–3.9 kJ m^{-1} s^{-1}	[666]
$ZrC_{\sim 1.0}$, sintered, porosity – 5%	Hollow cylinders: outer diameter – 48, wall thickness – 0.75	–	R = 50 K[e]	[336]
$ZrC_{\sim 1.0}$, sintered, porosity – 75%	Hollow cylinders: outer diameter – 52, wall thickness – 2.2	–	R = 68 K	[336]
$ZrC_{\sim 1.0}$	Linear dimension < 30 (area of the beam spot – 100 mm^2)	Thermal pulsed electron beam (energy – 10 keV) with power density – up to 18 kW cm^{-2} and pulse length – up to 1 s	Limiting thermal flux (experimetally determined for thermo-mechanical fracture, recalculated for the initial temperature of 20 °C and pulse length of 0.5 s) – 2.0 kW cm^{-2}	[1518]

[a]Figures-of-merit (criteria) of thermal shock/stress resistance [3, 4, 101, 737, 738]: $R = \sigma_t(1 - \nu)/\alpha E$, $R' = \sigma_t\lambda(1 - \nu)/\alpha E$, $R''' = E/\sigma_t^2(1 - \nu)$, $R'''' = \gamma E/\sigma_t^2(1 - \nu)$, where σ_t is tensile strength, ν is Poisson's ratio, α is thermal expansion coefficient, E is Young's modulus, λ is thermal conductivity and γ is fracture surface energy

[b]Values of maximal tangential stresses

[c]For the convenience of comparison the experimental values are normalized to the corresponding disk-shaped samples

[d]Values of maximal axis stresses

[e]*See also* Table 5.22

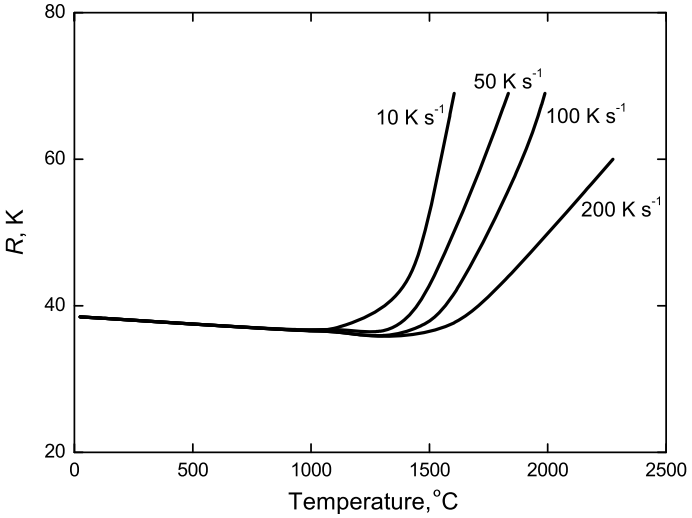

Fig. 5.35 Temperature dependence of thermal shock/stress resistance (thermal strength) R of sintered $ZrC_{\sim 1.0}$ materials at various heating rates [334, 335]

The variations of the elastic characteristics with higher pressure (up to 200 GPa) for quasi-stoichiometric zirconium monocarbide $ZrC_{\sim 1.0}$ based on *ab initio* calculations are shown in Fig. 5.34.

The experimental and calculated data on thermal shock/stress resistance (thermal strength) of single phase zirconium monocarbide ZrC_{1-x} materials are given in Table 5.18 and its temperature dependence – in Fig. 5.35.

The magnitudes of physico-mechanical (strength, elasticity) properties of near-stoichiometric zirconium monocarbide ZrC_{1-x} materials in the wide range of temperatures are summarized in Addendum in comparison with other ultra-high temperature materials.

5.5 Nuclear Physical Properties

Zirconium monocarbide ZrC_{1-x} is a material of special interest for the next-generation of nuclear engineering projects [74, 96, 798]; so, this trend makes its nuclear properties to be very important. The isotopes of elemental zirconium Zr (standard atomic mass – 91.224 u) from ^{78}Zr to ^{110}Zr, including metastable isomers (^{85m}Zr, ^{87m}Zr, ^{89m}Zr, $^{90m1-2}Zr$, ^{91m}Zr), and their general characteristics are summarized in Table 5.19; neutron nuclear physical properties of the isotopes of zirconium are given in Table 5.20. The comprehensive list of isotopes of element carbon C is presented in Table I-2.12 and its nuclear physical properties, including isotopic mass range, total number of isotopes, thermal neutron macroscopic cross sections, moderating ability and capture resonance integral, are given in Table I-A.8.

Table 5.19 General characteristics of the isotopes of zirconium [746–750]

Isotope	Mass, u	Abundance, %	Half-life period	Decay mode, excitation (radiation) energy, MeV
$^{90}Zr^{a}$	89.904704	51.45	–	–
$^{94}Zr^{a,b}$	93.906315	17.38	–	–
$^{92}Zr^{a,c}$	91.905041	17.15	–	–
$^{91}Zr^{a}$	90.905646	11.22	–	–
$^{96}Zr^{a,d}$	95.908273	2.80	2.0×10^{19} y	$\beta^-\beta^-$
$^{93}Zr^{e}$	92.906476	–	1.53×10^{6} y	β^-, 0.063
^{88}Zr	87.910227	–	83.4 d	*K*-capture; γ, 0.395
$^{95}Zr^{a}$	94.908043	–	64.032 d	β^-, 0.364 (54%), 0.396 (43%), 0.883 (3%); γ, 0.235, 0.722, 0.754
^{89}Zr	88.908890	–	3.267 d	β^+
^{97}Zr	96.910953	–	16.744 h	β^-, 1.91; γ, 0.75
^{86}Zr	85.91647	–	16.5 h	β^+
^{87}Zr	86.914816	–	1.68 h	β^+
^{84}Zr	83.92325	–	25.9 min	β^+
^{85}Zr	84.92147	–	7.86 min	β^+
^{89m}Zr	–	–	4.161 min	Isomer (93.77%), γ, 0.588; β^+ (6.23%), 0.85 (71%), 2.43 (29%)
^{83}Zr	82.92865	–	41.6 s	β^+ (>99.9%), β^+, *p* (<0.1%)
^{82}Zr	81.93109 (?)	–	32 s	β^+
^{98}Zr	97.912735	–	30.7 s	β^-
^{87m}Zr	–	–	14.0 s	Isomer, γ, 0.336
^{85m}Zr	–	–	10.9 s	Isomer (92%), γ, 0.292; β^+ (8%)
^{100}Zr	99.91776		7.1 s	β^-
^{81}Zr	80.93721	–	5.5 s	β^+ (>99.9%), β^+, *p* (<0.1%)
^{80}Zr	79.9404	–	4.6 s	β^+
^{102}Zr	101.92298	–	2.9 s	β^-
^{101}Zr	100.92114	–	2.3 s	β^-
^{99}Zr	98.916512	–	2.1 s	β^-
^{103}Zr	102.92660	–	1.3 s	β^-
^{104}Zr	103.92878 (?)	–	1.2 s	β^-
^{90m1}Zr	–	–	0.8 s	Isomer, γ, 2.319
^{105}Zr	104.93305 (?)	–	0.6 s	β^- (>99.9%), β^-, *n* (<0.1%)
^{106}Zr	105.93591 (?)	–	0.2 s (?)	β^-
^{107}Zr	106.94075 (?)	–	0.15 s (?)	β^-
^{108}Zr	107.94396 (?)	–	80 ms (?)	β^-
^{109}Zr	108.94924 (?)	–	60 ms (?)	β^-
^{79}Zr	78.94916 (?)	–	56 ms	β^+, *p*; β^+
^{78}Zr	77.95523 (?)	–	50 ms (?)	?
^{110}Zr	109.95287 (?)	–	30 ms (?)	?
^{91m}Zr	–	–	4.35 μs	Isomer, γ, 3.167
^{90m2}Zr	–	–	131 ns	Isomer, γ, 3.589

[a]Fission product
[b]Believed to decay by $\beta^-\beta^-$ to ^{94}Mo with a half-life over 1.1×10^{17} y
[c]Heaviest theoretically stable nuclide
[d]Primordial radionuclide, a half-life is longer than the age of the universe (nearly stable), theorized to also undergo β^- decay to ^{96}Nb
[e]Long-lived fission product

Table 5.20 Neutron nuclear physical properties of the isotopes of zirconium [751–754]

Isotope	Microscopic thermal neutron cross sections[a], b		Macroscopic thermal neutron cross sections[a], cm^{-1}		Resonance integral for neutron capture I_γ, b
	capture (absorption) σ_a	scattering[b] σ_s	capture (absorption) Σ_a	scattering[b] Σ_s	
Zr[c]	0.185 ± 0.003	6.46 ± 0.14	0.00775	0.340	0.93 ± 0.15
^{90}Zr	0.011 ± 0.005	5.10 ± 0.20	–	–	0.13 ± 0.02
^{94}Zr	0.050 ± 0.002	8.40 ± 0.40	–	–	0.27 ± 0.03
^{92}Zr	0.22 ± 0.06	6.90 ± 0.40	–	–	0.63 ± 0.02
^{91}Zr	1.24 ± 0.25	9.70 ± 0.20	–	–	5.20 ± 0.70
^{96}Zr	0.020 ± 0.001	3.80 ± 0.10	–	–	5.60 ± 0.20

[a]For 2200 m s^{-1} neutrons
[b]Total bound scattering cross sections
[c]Occurred naturally

The thermal neutron macroscopic cross sections Σ_i (*see* Eq. 2.53) of near-stoichiometric zirconium monocarbide ZrC_{1-x} (for 2200 m s^{-1} neutrons) [33]:

cross section of capture (absorption) Σ_a, cm^{-1}	7.27×10^{-3}
cross section of scattering Σ_s, cm^{-1}	~0.43

On the basis of the previous data on lattice dynamics and elasic constants [705] double differential (coherent elasic and incoherent inelastic) neutron scattering cross sections of the $ZrC_{\sim 1.0}$ crystal were calculted by Difilippo and Renier [790].

For the estimation of probable damage of zirconium monocarbide materials exposed to various types of radiation the parameters of formation and migration of lattice point defects (vacancies, antisites and interstitial atoms and clusters) are given in Table 5.21.

An example of such damage (swelling) produced by fast neutrons ($E > 1$ MeV) in the range of moderate temperatures is shown in Fig. 5.36. Severe fracturing of the sintered and hot-pressed materials is observed at moderate temperatures (130–355 °C) for fluences > (2.5–3.0) × 10^{21} cm^{-2} (dpa > 1.6) [758, 764]; with a fluence of 2.4 × 10^{21} cm^{-2} at elevated temperatures (1000–1100 °C) swelling was lower with relative expansion $\Delta V/V < 1\%$ [765, 775]. The severe fracturing is corresponding to the doses led to the maximal expansion of a ZrC_{1-x} phase lattice parameter [795]. The effect of irradiation with a fast neutrons fluence of 1.5 × 10^{20} cm^{-2} on $ZrC_{0.98}$ sintered materials at 150 and 1100 °C leads to the expansion of lattice parameter and decrease of density, which are not correlated with each other, as despite a lesser degree of lattice parameter expansion at higher temperatures, intergranular porosity increases and accelerates the process of swelling in general [538]. The behaviour of ZrC_{1-x} materials irradiated with fast neutrons ($E \geq 0.85$ MeV, $\Phi = 1.0 \times 10^{19}$ cm^{-2}, 140 °C) varies noticeably with a deviation from the stoichiometry, e.g. in irradiated $ZrC_{0.94}$ the relative expansion of lattice parameter is ~8 times higher than that of $ZrC_{0.73}$ treated at the same conditions; this difference in the behaviour, according to Andrievskii et al. [766], is connected with the possibillty of induced displacements of carbon atoms into tetrahedral sites, which exists in

Table 5.21 Parameters of formation and migration of lattice point defects (vacancies, antisites and interstitial atoms and clusters) in near-stoichiometric zirconium monocarbide[a]

Defect	Metal sublattice				Non-metal sublattice			
	E_f, eV	E_m, eV	S_f/k_B	S_m/k_B	E_f, eV	E_m, eV	S_f/k_B	S_m/k_B
Vacancy[b] [755, 756, 1419]	3.95	3.29	4.95	4.12	1.32	5.28	1.65	6.60
Vacancy[c] [756, 1419]	3.59	3.08	4.51	3.87	0.98	3.16	1.23	3.97
Vacancy[d] [735]	9.4	–	–	–	0.28	4.3[e]	–	–
Vacancy[f] [167]	8.89	–	–	–	1.16	–	–	–
Vacancy[g] [167]	5.80	–	–	–	1.51	4.86	–	–
Vacancy[h] [805]	8.92	–	–	–	1.20	–	–	–
Vacancy[i] [805]	8.81	–	–	–	0.95	–	–	–
Vacancy[i] [806]	8.83	–	–	–	0.93	–	–	–
Vacancy[i] [808]	8.89	–	–	–	0.93	–	–	–
Vacancy[i] [809]	7.19	5.44	–	–	0.93	4.41	–	–
Vacancy[j] [1287, 1401]	7.81	5.67	–	–	0.23	–	–	–
Vacancy[k] [1414]	4.72	–	–	–	–	–	–	–
Di-vacancy[j] [1287, 1401]	6.49	4.05[l]	–	–	6.49	4.05[l]	–	–
Di-vacancy[j] [1521]	6.97	–	–	–	6.97	–	–	–
Antisite[h] [805]	13.16	–	–	–	9.92	–	–	–
Antisite[i] [805]	12.97	–	–	–	9.65	–	–	–
Antisite[i] [806]	13.00	–	–	–	9.56	–	–	–
Interstitial atom[b] [755, 756]	19.14	0.66	24.06	0.83	5.94	0.66	7.47	0.83
Interstitial atom[i,m] [809]	10.36	0.47	–	–	3.56	0.27	–	–
Interstitial atom[i] [808]	17.69	–	–	–	4.54	–	–	–
Frenkel pairs[j] [1287, 1401]	18.61	–	–	–	5.27	–	–	–
Interstitial dimer cluster along <100>[h,n] [805]	–	–	–	–	4.43	–	–	–
Interstitial dimer cluster along <100>[i,n] [805]	–	–	–	–	4.39	–	–	–
Interstitial dimer cluster along <100>[i,n] [806]	10.66	–	–	–	4.43	–	–	–
Interstitial dimer cluster along <110>[h,o] [805]	–	–	–	–	4.54	–	–	–
Interstitial dimer cluster along <110>[i,o] [805]	–	–	–	–	4.52	–	–	–

(continued)

Table 5.21 (continued)

Defect	Metal sublattice				Non-metal sublattice			
	E_f, eV	E_m, eV	S_f/k_B	S_m/k_B	E_f, eV	E_m, eV	S_f/k_B	S_m/k_B
Interstitial dimer deviates 8° from <100>[h,p] [805]	–	–	–	–	4.27	–	–	–
Interstitial dimer deviates 8° from <100>[i,p] [805]	–	–	–	–	4.12	–	–	–
Interstitial trimer cluster along <101>[h,q] [805]	–	–	–	–	3.71	–	–	–
Interstitial trimer cluster along <101>[i,p] [805]	–	–	–	–	3.52	–	–	–
Interstitial trimer cluster along <101>[i,q] [806]	–	–	–	–	3.56	–	–	–
Interstitial trimer cluster (nonlinear)[h,r] [805]	–	–	–	–	3.66	–	–	–
Interstitial trimer cluster (nonlinear)[i,r] [805]	–	–	–	–	3.51	–	–	–
Interstitial trimer cluster (nonlinear)[i,s] [806]	–	–	–	–	3.56	–	–	–
Interstitial tetrahedron[h,t] [805]	–	–	–	–	3.98	–	–	–
Interstitial tetrahedron[i,t] [805]	–	–	–	–	3.78	–	–	–
Interstitial (along <111>) tetrahedron[i] [806]	8.72[u]	–	–	–	3.82[t]	–	–	–
Interstitial tetrahedron[i,v] [808]	8.80	–	–	–	3.61	–	–	–
Interstitial dimer cluster along <101>[i,w] [806]	9.96	–	–	–	–	–	–	–
Interstitial dimer cluster aligned to <111>[i,x] [806]	9.29	–	–	–	–	–	–	–
Interstitial dimer cluster aligned to <111>[i,y] [806]	9.24	–	–	–	–	–	–	–
V_{Zr}-$(V_C)_6$-cluster[j] [1287, 1401]	2.92	4.54[z]	–	–	2.92	4.54[z]	–	–

(continued)

Table 5.21 (continued)

[a]Denoted: E_f – defect formation energy, E_m – defect migration energy, S_f – defect formation entropy, S_m – defect migration entropy, k_B – Boltzmann constant
[b]Calculated on the basis of bonding model (relaxation displacement of atoms surrounding a defect is not taking into account)
[c]Calculated on the basis of elastic continuum model
[d]*Ab initio* simulation computed from the largest super-cell on the basis of the PAW pseudopotential using the PBE formulation
[e]Calculated value of activation energy for carbon self-diffusion in zirconium monocarbide E_A = 440 kJ mol^{-1} (*see also* Sect. 5.6, Table 5.29)
[f]Calculated using the Vienna *Ab initio* Simulation Package (VASP) [788, 789]
[g]Calculated using ultrasoft pseudopotential
[h]Calculated on the basis of the Vanderbilt-type ultrasoft pseudopotential with GGA exchange-correlation function
[i]Calculated on the basis of the projector-augmented plane-wave (PAW) method with PBE exchange-correlation function
[j]Calculated on the basis of PAW (using VASP) within GGA and nudged elastic band (NEB) methods
[k]Calculated on the basis of data on the yield strength determined by hardness and elasticity experimental measurements
[l]Calculated value of activation energy for metal self-diffusion in zirconium monocarbide (di-vacancy mechanism) E_A = 1020 kJ mol^{-1} (*see also* Sect. 5.6, Table 5.29)
[m]Defect formation and defect migration entropies are set to zero; calculated activation energies E_A of diffusion for defects: C vacancy – 515 kJ mol^{-1}, C interstitial – 370 kJ mol^{-1}, Zr vacancy – 1220 kJ mol^{-1} and Zr interstitial – 1050 kJ mol^{-1}
[n]Dumbbell structure cluster, interatomic distances: C–C – 0.124 nm, Zr–Zr – 0.176 nm
[o]Dumbbell structure cluster, interatomic distance C–C – 0.139 nm
[p]Dumbbell structure cluster, interatomic distance C–C – 0.125 nm
[q]Interatomic distances C–C–C – 0.139 nm
[r]Interatomic distances C–C – 0.146 nm with 139° angle between C–C pairs
[s]Interatomic distances C–C – 0.141 nm
[t]Interatomic distances C–C – 0.161 nm with 117° angle between C–C pairs
[u]Zr-centred double tetrahedron, interatomic distance Zr–Zr – 0.212 nm
[v]Calculated E_f of isolated Frenkel pairs for Zr atoms – 17.69 eV and C atoms – 4.54 eV and that of a Schottky defect – 8.14 eV
[w]Interatomic distance Zr–Zr – 0.184 nm
[x]Perturbed <100>, interatomic distance Zr–Zr – 0.1849 nm
[y]Perturbed <101>, interatomic distance Zr–Zr – 0.1854 nm
[z]Calculated value of activation energy for metal self-diffusion in zirconium monocarbide (cluster mechanism) E_A = 720 kJ mol^{-1} (*see also* Sect. 5.6, Table 5.29)

near-stoichiometric compositions, while in carbon-deficient compositions carbon atoms may be displaced to already vacant octahedral sites. Some data on the resistance of ZrC_{1-x} materials to fast neutron irradiation damage, based on the measurements of their physical properties after the irradiation, are summarized in Table 5.22; the data, including the results obtained by materials testing in reactor IVG-1 for modeling nuclear rocket engine reactor environment, show that the irradiation leads to appreciable changes of density and electrical resistivity accompanied by minor changes of strength, hardness and elastic characteristics [337, 761, 762]. The fast neutron irradiation ($E > 0.18$ MeV, $\Phi = 5 \times 10^{21}$ cm^{-2}, 1200 °C) behaviour of $ZrC_{\sim 1.0}$ chemically vapour deposited coatings (thickness – 45–52 μm) on the

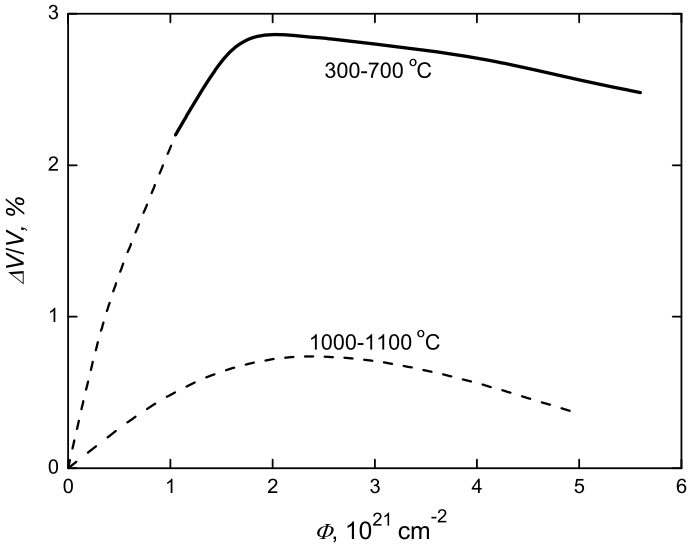

Fig. 5.36 The change in macroscopic volume (swelling) with fast neutron (>1 MeV) fluence Φ for produced by ceramic technologies (hot-pressed, slip-cast, explosion-pressed) near-stoichiometric zirconium monocarbide ZrC_{1-x} materials irradiated at different temperatures [755, 758, 764, 765, 1519]

Table 5.22 Structural, electrical and physico-mechanical properties of $ZrC_{0.98}$ sintered materials before and after fast neutron irradiation ($\Phi = 1.5 \times 10^{20}$ cm^{-2}) exposure at different temperatures [4, 337, 538]

Characteristics	Unit	Measured values		
		Initial samples 20 °C	Irradiated samples[a]	
			150 °C	1100 °C
Density	g cm^{-3}	6.40	6.29	6.27
Lattice parameter	nm	0.4692	0.4714	0.4698
Electrical resistivity	μΩ m	0.43	2.50	0.65
Flexural strength:				
average	MPa	245	315	255
minimal	MPa	215	235	235
maximal	MPa	300	340	290
Young's modulus	GPa	400	405	400
Microhardness	GPa	20.1	22.5	21.6
load[b]	N	0.98	0.39	–
Thermal stress resistance criterium R[c]	K	45	75	75

[a]The neutron flux – 10^{12}–10^{15} cm^{-2} s^{-1}
[b]The minimum load during the microhardness tests, at which all the indentations were cracking
[c]*See* notes to Table 5.18

experimental fuel particles was studied by Reynolds et al. [767]; at something similar irradiation conditions, but at rather higher temperatures Fukuda et al. ($E > 0.18$ MeV, $\Phi = 1.4 \times 10^{21}$ cm^{-2}, 1260 °C) [781], Ogawa et al. ($E > 0.18$ MeV, $\Phi = (1.0–2.2) \times 10^{21}$ cm^{-2}, 900–1600 °C) [768, 782–785] and Minato et al. ($E > 0.18$ MeV, $\Phi = 1.2 \times 10^{21}$ cm^{-2}, 1400–1650 °C) [769–773] also confirmed good mechanical and environmental stability of $ZrC_{\sim 1.0}$ coatings on the fuel particles. High purity zone refined $ZrC_{0.93}$ materials were fast neutron irradiated by Snead et al. ($E > 0.1$ MeV, $\Phi = (1.0–10.0) \times 10^{21}$ cm^{-2}, 640–1500 °C) [681], measurements of lattice parameter indicated a lack of significant crystalline strain due to irradiation, only modest changes in the physico-mechanical properties of fracture toughness, hardness and elastic modulus and slight reduction of thermal conductivity near 1000 °C irradiation (though the parameter was essentially unchaged at higher temperatures) were observed; so, the general finding of the studies was that the materials are quite stable under neutron irradiation in the testing conditions.

The damage produced by thermal neutrons fluences up to 1.5×10^{20} cm^{-2} at ~50 °C results in the slight failure of hot-pressed $ZrC_{0.98}$ materials, which is accompanied by ~0.3% relative change of lattice parameter, but ~190% relative change of specific electrical resistance (Table 5.23), while the accumulated energy in the materials corresponding to these irradiation conditions amounts to 12.7 kJ mol^{-1} [755, 759, 760, 794]. By *in situ* property measurements Taubin et al. [774] found out only minor changes in heat capacity, thermal conductivity, thermal expansion and electrical resistivity in $ZrC_{0.96}$ irradiated below 400 °C with mixed fission neutrons ($\Phi = \sim 1.0 \times 10^{16}$ cm^{-2}). According to Markin et al. [786] the damage ability of monocarbide ZrC_{1-x} materials under the action of neutron irradiation in the part of magnetic susceptibility only weakly depends on the deviation from the

Table 5.23 The effect of thermal neutron irradiation on the structure and physical properties of zirconium monocarbide ZrC_{1-x} hot-pressed materials [755, 759, 760, 763, 794, 1520]

Characteristics	Unit	Temperature, °C	Initial values	After irradiation with different thermal neutron fluences Φ, 10^{19} cm^{-2}			
				1.0	3.7	7.5	15
			$ZrC_{\sim 1.0}$				
Flexural strength	MPa	80–100	200	158	155	153	–
			$ZrC_{0.98}$				
Lattice parameter	nm	~50	0.4679	0.4681	0.4683	0.4689	0.4694
Electrical resistivity	μΩ m	~50	0.56	0.68	0.95	1.37	1.63
relative change	%	~50	0	+23	+70	+145	+191
Microhardness	GPa	~50	26.4	26.0	25.5	27.2	30.1
relative change	%	~50	0	−1.5	−3.4	+3.0	+14
Accumulated energy[a]							
per gram	J g^{-1}	~50	0	18	32	75	123
per mole	kJ mol^{-1}	~50	0	1.86	3.34	7.70	12.7
Effective temperature[a]	K	~50	–	80	140	310	490

[a]For additional information

stoichiometry of monocarbide phase. The effect of earlier stages (< 30 s) of nuclear reactor irradiation with neutron flux $\varphi = \sim 1.0 \times 10^{14}$ cm^{-2} s^{-1} and gamma quanta dose of 2.58×10^6 C kg^{-1} on the heat capacity, thermal conductivity and electrical resistivity of $ZrC_{\sim 1.0}$ materials, when the gamma quanta influence dominates, was studied by Taubin [787]. A study of the primary damage due to neutron irradiation created by a high energy collision cascade (50 keV) in single crystal stoichiometric zirconium monocarbide was conducted by Van Brutzel and Crocombette [791] by means of classical molecular dynamics simulation.

To simulate high irradiation damage rate in near-stoichiometric zirconium monocarbide Gan et al. [776] conducted Kr ion irradiation (E = 1 MeV, $\Phi = 2.5 \times 10^{15} - 1.75 \times 10^{16}$ cm^{-2}, corresponding to 10–70 dpa) of hot-pressed $ZrC_{\sim 1.0}$ materials (porosity < 1%, contents: O – 0.21%, N – 0.61%, Hf – 1.90%, Ti – 0.19%) at room temperature (10 and 30 dpa) and 800 °C (10 and 70 dpa) with a damage rate $\sim 3.0 \times 10^{-3}$ dpa s^{-1}; as a result no radiation induced amorphization was found, and the maximal lattice expansion of ~7% was observed at higher irradiation intensity and higher temperature. The same $ZrC_{\sim 1.0}$ materials were also irradiated with protons (E = 2.6 MeV, $\Phi = 2.75 \times 10^{19}$ cm^{-2}, corresponding to 0.7–1.8 dpa) at 800 °C [777]; in comparison to Kr ion irradiation (10 dpa) mentioned above, proton irradiated materials show less damage to the lattice structure and exhibit faulted dislocation loops on (111) planes, which were not observed in the Kr ion irradiated materials. To irradiate the same materials to doses 0.7 and 1.5 dpa, Yang et al. [778, 792] also used a 2.6 MeV proton beam at 600–900 °C to doses up to 1.75 dpa and found out the lattice expansion at 600–800 °C and slight lattice contraction at 900 °C, which were in good agreement with the earlier reported results from neutron irradiation; the hardness and the fracture toughness of materials both increase with increasing the radiation dose: for 1.5 dpa dose the relative increases were 10–15 and 50–100%, respectively. Jensen et al. [341–342, 799–800] reported about ~66% degradation of thermal conductivity (up to 11.9 W m^{-1} K^{-1}) in the ~50 μm deep zone of hot-pressed $ZrC_{\sim 1.0}$ materials proton-irradiated (E = 2.6 MeV, 1.75 dpa, 600 °C). In the range of doses from 1 to 3 dpa in the zone refined ZrC_{1-x} materials with $0 < x \leq 0.1$ and non-combined C content – up to 2.3%, which were irradiated with protons (E = 2.0 MeV, $\varphi = 1.0 \times 10^{18}$ cm^{-2} s^{-1}, dose rate at the depth of 5–20 μm – 4.7×10^{-5} dpa s^{-1}, 800 ± 10 °C), the following parameters of irradiation-induced defects were characterized and estimated: average size of dislocation loops – from 7 to 28 nm and density of dislocation loops from 1.5×10^{15} cm^{-3} to 7.5×10^{15} cm^{-3} [803]. In the similar materials ($0 < x \leq 0.16$ and non-combined C content – up to 1.9%) irradiated with protons (E = 2.0 MeV, $\Phi = 2.0 \times 10^{19}$ cm^{-2} corresponding to ~2 dpa for $ZrC_{1.0}$, dose rate at the depth of 10–15 μm – 1.8×10^{-5} dpa s^{-1}, 1125 °C) Yang et al. [804] identified the dislocation loops as interstitial type perfect loops with average size from 9.85 nm (for $ZrC_{0.95}$) to 53 nm (for $ZrC_{\sim 1.0}$ with content non-combined

C – 1.9%); it was indirectly proved that the dislocation loop core likely consists of carbon atoms, and graphite precipitates in the super-stoichiometric compositions are detrimental, as the dramatic increases in the size of and density of dislocation loops were observed in the vicinity of these precipitates in monocarbide phases. Proton irradiation induced faceted voids were only observed in a near-stoichiometric composition, which is attributed to the pre-existing dislocation lines as biased sinks for vacancies.

Sintered $ZrC_{\sim 1.0}$ materials were irradiated with fast and energetic Kr ions (E = 25.8 MeV, Φ = (1.0–6.0) × 10^{16} cm^{-2}); the study considered two domains produced by the inelastic and the elastic collisions that occurred during the process of ion irradiation, significant thermal conductivity degradation in both domains was observed, elastic collisions were shown to degrade the thermal properties of $ZrC_{\sim 1.0}$ more strongly, reducing thermal conductivity from 20 W m^{-1} K^{-1} before irradiation to less than 1 W m^{-1} K^{-1} [331, 332]. Floating-zone refined $ZrC_{0.80}$ and $ZrC_{0.90}$ were also irradiated with Kr ions (E = 1 MeV, φ = 6.25 × 10^{11} cm^{-2} s^{-1}, maximal dose from 3.9 to 12.8 dpa, −253 to 800 °C), no long-range migration of the visible defects or dynamic defect creation and elimination were observed during the irradiation, but some coarsening of the microstructure with the formation of dislocation loops was observed at higher temperatures; the irradiated microstructure was found to be only weakly dependent on the stoichiometry [807]. Gosset et al. [779, 780] performed low energy Au ions (Φ = 1.0 × 10^{12} – 5.0 × 10^{15} cm^{-2}, corresponding to 7–35 dpa) irradiation of hot-pressed $ZrC_{0.95}$ and $ZrC_{0.85}O_{0.08}$ at room temperature, high sensitivity to oxidation was observed with the formation of oxide precipitates during the irradiation process; three stages of damage could be classified: at lower fluences (Φ < 1.0 × 10^{12} cm^{-2}) – low oxide modifications (*see also* Table 5.26) are formed, at intermediate fluences (1.0 × 10^{12} cm^{-2} < Φ < 1.0 × 10^{14} cm^{-2}) – high micro-strains appear jointly with small faulted dislocation loops and at higher fluences (Φ > 1.0 × 10^{14} cm^{-2}) – the micro-strains saturate and the loops coalesce to form a dense dislocation network. The materials showed the moderate lattice increase up to 0.6 vol.%, which saturates at Φ = ∼1.0 × 10^{14} cm^{-2}. No amorphization was observed in (001)-oriented single crystal $ZrC_{\sim 0.9}$ (99.99% purity) irradiated with Au ions beam (E = 1.2 MeV, Φ = 3.0 × 10^{13} – 3.0 × 10^{16} cm^{-2}, 25 °C) as well, with increasing fluence the irradiation damage was found to spread in the materials deeper than the depth of direct damaging by the ion beam [801, 802, 810, 929]. Jiang et al. [811] showed that the metal sublattice in zirconium monocarbide remains structurally stable against carbon sublattice displacements up to the highest dose of 1.0 dpa; consequently, ZrC_{1-x} cannot be amorphized by the accumulations of Frenkel pairs of carbon atoms like it is occurred with covalent carbides. A comparative study of ZrC_{1-x} materials with micro- and nano-sized grains irradiated with low energy heavy ions concluded that the resulting damaged surface area of the materials has a 0.1–0.2 μm thickness, both

materials with micro- and nano-sized grains showed high internal distortions, for micro-sized-grain materials a linear volume swelling in the range of 20–40 Zr dpa was observed [180]. Spark-plasma sintered $ZrC_{0.95}O_{0.05}$ materials were implanted with Xe ions (E = 0.8 MeV, $\Phi = 1.0 \times 10^{15}$ and $\Phi = 1.0 \times 10^{16}$ cm^{-2}, 25 °C), in the as-implanted samples Xe bubles were not formed at both fluences and the point defects as vacancies, interstitials and Xe-substitutions were predominant, the higher fluence led to thinner damaged layer; the post-implantation annealing of the samples treated with higher fluence caused formation of Xe bubbles, and the concentration of bubbles increased with the time of the high vacuum ($5 \times 10^{-4} - 5 \times 10^{-5}$ Pa) annealing at 1800 °C indicating good Xe retention, while the annealing of samples treated with lower fluence at first caused clustering of defects, and only after more than 8 h of annealing led to the formation of Xe bubbles [793]. It was also shown that the presence of residual oxide inclusions (~2%) and some free carbon (~0.15%) induces a plate-like damaging of the materials surface whereas pure Xe-implanted $ZrC_{0.95}O_{0.05}$ materials display a less damaged surface after vacuum annealing procedure [796]. First principles calculations were employed to study the diffusion behaviour of He impurities in ZrC_{1-x} materials; it was revealed that Zr-vacancy assisted mechanism plays a dominant role in it with a small energy barrier of 0.70 eV. He is likely trapped in Zr pre-existing vacancies, which may impact on the mechanical properties and dimensional stability of ZrC_{1-x} materials [808].

The details of radiochemical compatibility of ZrC_{1-x} materials with some fission products and diffusion parameters of these products in ZrC_{1-x} phases are given in Table 5.27. Nuclear physical properties of zirconium monocarbide ZrC_{1-x} in comparison with other ultra-high temperature materials are also given in Addendum.

5.6 Chemical Properties and Materials Design

The comprehensive data on the chemical properties, compatibility (in the connection with both environmental resistance and composite materials design) and interaction behaviour of zirconium monocarbide at elevated, high and ultra-high temperatures with elements (metals, non-metals) are summarized in Table 5.24, with refractory compounds – in Table 5.25 and with gaseous media – in Table 5.26. The data on the oxidation resistance of zirconium monocarbide materials listed there are also accompanied by the graphic information in Figs. 5.37 and 5.38; the isothermal oxidation kinetics of near-stoichiometric ZrC_{1-x} materials can be considered in the context of ridge-effect model proposed by Shabalin [984, 985] with ridge temperatures at lower oxygen pressures: nearby ~900 °C – for denser materials (with lower porosity) and around ~550 °C – for highly porous (or more active chemically) materials. The details of radiochemical compatibility of ZrC_{1-x} materials with some fission products and diffusion parameters of these products in

Table 5.24 Chemical interaction and/or compatibility of zirconium monocarbide with elements (metals, non-metals) at elevated, high and ultra-high temperatures, including solid matters and molten media (reaction systems are given mainly in alphabetical order)[a]

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	Reference
ZrC_{1-x}–Ag			*See* Tables 5.27 and 5.28	
ZrC_{1-x}–Ag–Pd	–	1230	Formation of $ZrPd_3$	[1014]
ZrC_{1-x}–Al	–	<1000	No chemical interaction; the partial dissolution of Zr from ZrC_{1-x} into Al melt was observed	[1, 30, 32, 33, 42, 45, 48, 63, 812–839, 906, 907, 911–913, 934, 1011, 1025, 1026, 1071, 1082, 1090, 1091, 1233, 1366, 1375, 1404]
	–	1000	The contact interaction is weak (5 h exposure)	
	Ar	1000–1250	Weak interaction during 0.2 h exposure	
	He	1000–1400	Formation of Al_4C_3	
	–	≥1250	Intensive interaction and formation of Al solid solutions in carbide	
	–	1700	Intensive interaction with Al vapour and formation of new refractory phases	
	–	–	Formation of $Zr_3Al_3C_5$, $Zr_3Al_4C_6$, $Zr_2Al_3C_5$, Zr_5Al_3C, $Zr_2Al_3C_4$, $Zr_2Al_4C_5$, $ZrAl_8C_7$, $ZrAl_3C_3$, $ZrAl_4C_4$, $ZrAlC_{2-x}$, Zr_3AlC_2 and Zr_2AlC	
	–	–	Hexagonal ZrC_{1-x} platelets are formed via self-propagating high-temperature synthesis from Al–Zr–C powder mixtures	
	–	–	The effect of Al as an additive for hot-pressing process was studied	
			Data on chemical resistance in molten Al available in literature are controversial	
			See also section ZrC_{1-x}–Al_4C_3 in Table 5.25	
			See also Table 5.28	
			See also section C–Al–Zr in Table I-2.14	
ZrC_{1-x}–Al–Cu	Vacuum	550	Particles of ZrC_{1-x} (0.3–6 μm) are compatible with Al – 4 mas.% Cu matrix alloy	[946]
ZrC_{1-x}–Al–Mg	–	750	Particles of ZrC_{1-x} (1–3 μm) show high stability and uniform distribution in molten Al – 1% Mg alloy, even though minute amounts of Al_3Zr were observed	[1044]
ZrC_{1-x}–Al–Zr	–	–	Formation of $Zr_3Al_3C_5$, $Zr_3Al_4C_6$, $Zr_2Al_3C_5$, Zr_5Al_3C, $Zr_2Al_3C_4$, $Zr_2Al_4C_5$, $ZrAl_8C_7$, $ZrAl_3C_3$, $ZrAl_4C_4$, $ZrAlC_{2-x}$, Zr_3AlC_2 and Zr_2AlC	[814–825, 828, 829, 834–839, 906, 907, 911–913, 1011, 1090, 1091, 1366, 1404]

(continued)

Table 5.24 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	Reference
ZrC_{1-x}–Al_2O_3–Al	–	–	Compatible as components of metal matrix composites	[1363]
ZrC_{1-x}–NbC_{1-x}–Al			*See* section NbC_{1-x}–ZrC_{1-x}–Al in Table 4.18 *See also* section C–Al–Nb–Zr in Table I-2.14	
ZrC_{1-x}–SiC–Al	–	–	Formation of $Zr(Al_{1-x}Si_x)_8C_7$, $Zr(Al_{1-x}Si_x)_4C_4$, $Zr_2(Al_{1-x}Si_x)_4C_5$ and $Zr_3(Al_{1-x}Si_x)_4C_6$ *See also* section ZrC_{1-x}–Al_4C_3–Si in this table *See also* section ZrC_{1-x}–Al_4C_3–SiC in Table 5.25	[826, 827, 831–833, 904, 1082]
ZrC_{1-x}–SiC–TiC_{1-x}–Al	–	–	Formation of $(Ti,Zr)_3(Si,Al)C_2$ ($M_{n+1}AX_n$-phase solid solution) *See also* section ZrC_{1-x}–Al_4C_3–SiC–TiC_{1-x} in Table 5.25	[834, 838]
ZrC_{1-x}–β-B	– –	1500 2000–2600	Solubility of B in ZrC_{1-x} is very low Formation of $ZrB_{2\pm x}$ *See also* section C–B–Zr in Table I-2.14	[1, 33, 46, 76, 200, 840–845, 914, 1030, 1403, 1462]
ZrC_{1-x}–$B_{4\pm x}C$–β-B	Chamber of anvil type	1600–2000 (0.015–5 GPa)	Formation of $ZrB_{2\pm x}$; the addition of B favours the reaction between ZrC_{1-x} and $B_{4\pm x}C$ going to completion *See also* section C–B–Zr in Table I-2.14	[1030, 1462]
ZrC_{1-x}–Ba			*See* Table 5.27	
ZrC_{1-x}–Be	–	–	The effect of Be as an additive for hot-pressing process was studied	[33]
ZrC_{1-x}–Bi	Ar Air	280–320 375	No chemical interaction during 10 h exposure Traces of Zr were revealed in liquid metal after 10 h exposure *See also* Table 5.28	[1, 30, 32, 33, 42, 45, 48, 63]
ZrC_{1-x}–Bi–Pb	–	1100	Testing materials (content non-combined C – 4.1%) in dynamic conditions (24 h exposure) showed disintegration rate – 5.25 μm h^{-1}, pitting type of attack was observed	[33, 846]
ZrC_{1-x}–C	– –	<2750 ~2850–2930	ZrC_{1-x} is compatible in contact with any C (graphite) parts and articles Eutectic ZrC_{1-x}–C (graphite) *See also* section C–Zr in Table I-2.13	[1, 30, 33, 99, 102, 559, 841, 847, 848, 901, 920, 995, 1049, 1089, 1395, 1534]

(continued)

Table 5.24 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	Reference
ZrC_{1-x}–TiC_{1-x}–δ-$WC_{1\pm x}$–C	–	1300–1800	Isothermal sections with miscibility gaps determined	[1074]
ZrC_{1-x}–VC_{1-x}–δ-$WC_{1\pm x}$–C	–	1300–1500	Isothermal sections with miscibility gaps determined	[1074]
ZrC_{1-x}–C–Fe			*See* Table 5.28	
ZrC_{1-x}–C–Fe–Mn			*See* Table 5.28	
ZrC_{1-x}–C–Ni			*See* Table 5.28	
ZrC_{1-x}–Cd	Air	450	The contact interaction is very weak: after 10 h exposure the content of Zr in the melt achieved 0.01%	[30, 32, 33, 42, 45, 48, 63]
ZrC_{1-x}–Ce			*See* Table 5.27	
ZrC_{1-x}–α/ε-Co	–	800	ZrC_{1-x} is in equilibrium with all Zr intermetallides, α-Zr and α-Co; the solid solubility of Co in ZrC_{1-x} is negligible	[1, 4, 32, 33, 42, 45, 48, 63, 200, 849–855, 901, 1226, 1230, 1233]
	–	~1340–1410	Eutectic ZrC_{1-x}–α-Co, the mutual solubilities between the components are low; the eutectic structure is plate-like mainly	
	–	1500	Liquid metal penetrates into bulk ZrC_{1-x} (two-phase contact zone) with formation of solid solutions; the contact interaction is very intensive (5 h exposure), mutual exchange (dissolution) between the solid and liquid phases was observed	
	N_2 + CO	1550	Noticeable interaction between the components (0.2 h exposure)	
	–	>1550	ZrC_{1-x} materials are corroded by liquid Co and Co containing melts	
			See also Table 5.28	
			See also section C–Co–Zr in Table I-2.14	
ZrC_{1-x}–Co–Ni			*See* Table 5.28	
ZrC_{1-x}–Co–Ni–W			*See* Table 5.28	
ZrC_{1-x}–HfC_{1-x}–Co			*See* section HfC_{1-x}–ZrC_{1-x}–Co in Table 3.19	
			See also section C–Co–Hf–Zr in Table I-2.14	
ZrC_{1-x}–NbC_{1-x}–Co			*See* section NbC_{1-x}–ZrC_{1-x}–Co in Table 4.18	
			See also section C–Co–Nb–Zr in Table I-2.14	
ZrC_{1-x}–TaC_{1-x}–Co			*See* section TaC_{1-x}–ZrC_{1-x}–Co in Table 2.21	
			See also section C–Co–Ta–Zr in Table I-2.14	
ZrC_{1-x}–TiC_{1-x}–Co	–	1320	Eutectic $(Zr,Ti)C_{1-x}$ (x = 0.19)–$(Ti,Zr)C_{1-x}$ (x = 0.20)–α-Co	[855, 1074]
			See also section C–Co–Ti–Zr in Table I-2.14	

(continued)

Table 5.24 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	Reference
ZrC_{1-x}–TiC_{1-x}–VC_{1-x}–δ-$WC_{1\pm x}$–Co	–	1200–1800	General consideration of the system	[1074]
ZrC_{1-x}–TiC_{1-x}–δ-$WC_{1\pm x}$–Co	–	1300–1500	General consideration of the system; a miscibility gap in the ZrC_{1-x}–TiC_{1-x} system in the presence of Co is determined; spinodal decomposition during the sintering process of alloys with Zr/Ti ratios 3:7 and 5:5 and 10–30 mol.% of W carbide was revealed	[1074, 1375, 1526]
ZrC_{1-x}–VC_{1-x}–Co	–	1300	Eutectic $(Zr,V)C_{1-x}$ (x = 0.19)–$(V,Zr)C_{1-x}$ (x = 0.17)–α-Co *See also* section C–Co–V–Zr in Table I-2.14	[855, 1074]
ZrC_{1-x}–VC_{1-x}–δ-$WC_{1\pm x}$–Co	–	1300–1500	General consideration of the system; a miscibility gap in the ZrC_{1-x}–VC_{1-x} system in the presence of Co is determined	[1074]
ZrC_{1-x}–Cr	–	1120–1140 (?)	Eutectic ZrC_{1-x} (x = 0.03)–Cr; the max. solubility of Cr in ZrC_{1-x} is 2.3–8.5 mol.% (?) and that of ZrC_{1-x} in Cr is $\leq$0.1 mol.%	[1, 4, 30, 32, 33, 42, 45, 48, 63, 200, 856–864, 901, 995, 1059, 1429, 1523]
	Vacuum, 10^{-3} Pa	1200–1300	Almost no interaction between the components was observed during 100 h exposure	
	Vacuum, 10^{-3} Pa	1400	Formation of $Cr_{23}C_{6\pm x}$ (the impurities contained in metallic Cr were found to affect the interaction in the system considerably)	
	Vacuum, Ar	1730–1805	Eutectic ZrC_{1-x} (x = 0.10)–Cr; the max. solubility of Cr in ZrC_{1-x} is ~1.5–3.0 mol.% and that of ZrC_{1-x} in Cr is 1.3 mol.%; ZrC_{1-x} is in equilibrium with all Cr carbides and intermetallide and metallic Cr	
	–	–	Stable and compatible with each other as cermet components	
	–	1830	The contact interaction is intensive (5 h exposure)	
	Ar (pure)	~1860	ZrC_{1-x} is resistant to Cr melts (contact zone thickness <0.5 mm, at long-term exposures)	
	N_2 + CO	1900	No chemical interaction during 0.2 h exposure	
	–	>1900	ZrC_{1-x} materials are corroded by liquid Cr and Cr containing melts Data available in literature are controversial *See also* section C–Cr–Zr in Table I-2.14	
ZrC_{1-x}–Cr–Cu–Fe			*See* Table 5.28	

(continued)

Table 5.24 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	Reference
ZrC_{1-x}–Cr–Fe			*See* Table 5.28	
ZrC_{1-x}–Cr–Ni			*See* Table 5.28	
ZrC_{1-x}–$Cr_{23}C_{6\pm x}$–Cr	–	1575	Eutectic ZrC_{1-x}–$Cr_{23}C_{6\pm x}$–Cr *See also* section ZrC_{1-x}–$Cr_{23}C_{6\pm x}$ in Table 5.25 *See also* section C–Cr–Zr in Table I-2.14	[860–863, 917–919]
ZrC_{1-x}–Cs	Cs vapour	1100–1800	Solubility of Cs in ZrC_{1-x} decreases with increasing temperature from ~5×10^{-2} mol.% at 1100 °C to ~7×10^{-5} mol.% at 1800 °C	[33, 96, 489, 495, 500, 504, 865–867]
	Cs vapour	~1300–1700	Only adsorption interaction was observed on the surface of carbide powder (exposure for many hours)	
	Cs vapour	1400	Surface carbide composition unchanged after many hours of heating exposure	
	Cs vapour, ~13 Pa	~2000	No effect upon structural characteristics of ZrC_{1-x} and its thermionic emission parameters (some tens of hours exposure), adsorption character of interaction	
	Cs vapour	–	Chemical reaction with deposited by vacuum evaporation carbide layers (composition and structure uncontrolled) is reported	
	–	–	The interaction with Cs as a fission product and O released during fission is predicted to form Cs_2ZrO_3 *See also* Table 5.27 *See also* section C–Cs–Zr in Table I-2.14	
ZrC_{1-x}–Cu	Vacuum	300–800	Nanocrystalline ZrC_{1-x} films (mean grain size <5 nm) grown on Si(100) thermally stable in contact with Cu (0.5 h exposure)	[1, 33, 48, 850, 868, 944, 989, 1060, 1233, 1527]
	–	1100	Weak contact interaction (5 h exposure); the partial dissolution of ZrC_{1-x} in Cu melt was observed	
	–	>1100	ZrC_{1-x} materials migrate into molten Cu	
	Vacuum, 0.3 Pa	1100–1300	Formation of $Zr_xCu_yC_z$ (?) in the presence of W and/or δ-$WC_{1\pm x}$	
	Ar	<1150	No chemical interaction *See also* Table 5.28 *See also* section C–Cu–Zr in Table I-2.14	
ZrC_{1-x}–Cu–Fe–Si			*See* Table 5.28	
ZrC_{1-x}–Cu–Ni			*See* Table 5.28	
ZrC_{1-x}–Cu–V–Zr			*See* Table 5.28	
ZrC_{1-x}–Cu–Zr			*See* Table 5.28	
ZrC_{1-x}–Eu			*See* Table 5.27	

(continued)

Table 5.24 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	Reference
ZrC_{1-x}–α/γ/δ-Fe	–	1100	Formation of $Zr_3Fe_3C_{1-x}$, solubility of ZrC_{1-x} in γ-Fe is $\leq$0.5 mol.%; ZrC_{1-x} is in equilibrium with Fe and Fe–Zr intermetallides	[1, 4, 32, 33, 48, 63, 168, 200, 850, 869, 873, 901, 1025, 1026, 1076, 1226, 1230, 1233, 1439, 1531]
	–	1200	Formation of a new phase in the contact zone between ZrC_{1-x} powder and compact Fe	
	–	1400	Formation of a liquid phase in the contact zone between ZrC_{1-x} powder and compact Fe	
	–	~1380–1480	Eutectic ZrC_{1-x}–δ-Fe, the max. solid solubility of ZrC_{1-x} in δ-Fe is $\leq$~1 mol.%; eutectics structure is the complex system of colonies, which consists of the carbide plates (less rarely – filaments) located in the metallic matrix, in the hypereutectic alloys primary carbide crystals have predominantly cubic forms (along with them dendrite crystals are also observed)	
	Vacuum	1550	Liquid Fe penetrates into the bulk dense carbide (depth of penetration – ~300 μm) and forms solid solutions (in the contact zone carbide has a changed lattice parameter); mutual exchange (dissolution) between the solid and liquid phases was observed	
	Ar	1550	Liquid Fe penetrates into the bulk dense carbide (depth of penetration – ~40 μm) and forms solid solutions	
	–	>1550	ZrC_{1-x} materials are corroded by liquid Fe and Fe containing melts	
	–	1560	The contact interaction is very intensive (5 h exposure)	
	–	–	According to DFT calculations the $ZrC_{\sim 1.0}$ (100)/ Fe (110) interface is well lattice-matched (~1.7%) producing a fairly smooth interface with little structural relaxation; semicoherent interfacial energy at relaxed interface – 0.66 J m^{-2} *See also* Table 5.28 *See also* section C–Fe–Zr in Table I-2.14	
ZrC_{1-x}–Fe–Ni			*See* Table 5.28	
ZrC_{1-x}–Fe–Si			*See* Table 5.28	
ZrC_{1-x}–Ga	Ar	<800	No chemical interaction *See also* Table 5.28	[1, 850]
ZrC_{1-x}–Ge	Ar	<1000	No chemical interaction *See also* Table 5.28	[1, 850]
ZrC_{1-x}–Al_4C_3–Ge	–	–	Formation of $Zr_2Al_3(Al,Ge)C_5$ and $Zr_3Al_3(Al,Ge)C_6$ (intergrowth structures, consisting of Zr_mC_{m+1} (NaCl-type) slabs separated by $(Al_{4-z}Ge_z)C_4$ (Al_4C_3-type) layers)	[905]

(continued)

Table 5.24 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	Reference
ZrC_{1-x}–Hf			*See* section C–Hf–Zr in Table I-2.14	
ZrC_{1-x}–In	Ar	<250	No chemical interaction	[1, 836–839, 850, 1530]
	–	–	Formation of Zr_2InC ($M_{n+1}AX_n$-phase) *See also* Table 5.28 *See also* section C–In–Zr in Table I-2.14	
ZrC_{1-x}–TiC_{1-x}–In	–	–	Formation of $(Zr,Ti)_2InC$ ($M_{n+1}AX_n$-phase solid solution); $(Zr_{0.5}Ti_{0.5})_2InC$ was synthesized	[837, 838, 1404, 1405]
ZrC_{1-x}–Ir	–	–	In the reaction with metallic Ir carbide is decomposed to form graphite and intermetallide $ZrIr_{3-x}$ (with defective ordered structure)	[200, 990, 991, 1374]
	–	1500–2000	ZrC_{1-x} is in equilibrium with $ZrIr_{3\pm x}$ and (Zr,Ir) alloys	
	–	~2030–2130	Eutectic (estimated) ZrC_{1-x} ($x \approx 0.18$)–Ir *See also* section C–Ir–Zr in Table I-2.14	
ZrC_{1-x}–K	–	~2000	No chemical reaction observed, no effect upon structural characteristics of ZrC_{1-x} and its thermionic emission parameters (some tens of hours exposure)	[504, 866, 1031, 1052]
ZrC_{1-x}–$ZrB_{2\pm x}$–La	Vacuum	~1800	The addition of La was found to be a very effective aid for spark-plasma sintering, but not useful for the conventional sintering process, because of the intensive evaporation of La in those conditions	[1453, 1457]
ZrC_{1-x}–Li	–	–	The results of screening tests on ZrC_{1-x} for molten Li are given	[1024, 1025]
ZrC_{1-x}–Mg	–	>650	ZrC_{1-x} materials migrate into Mg melts	[48, 63, 989, 995]
	–	700–800	Weak contact interaction (2 h exposure)	
	–	>800	ZrC_{1-x} materials were corroded intensively by liquid Mg and Mg containing melts	
ZrC_{1-x} – $\alpha/\beta/\gamma/\delta$-Mn	Vacuum, Ar	1250	The mutual exchange (dissolution) between the solid (ZrC_{1-x}) and liquid (Mn) phases was observed in the contact zone	[1, 33, 48, 63, 850, 1233]
	–	1270–1300	Intensive contact interaction with formation of eutectic alloys and solid solutions (5 h exposure)	
	–	>1300	ZrC_{1-x} materials are corroded by liquid Mn and Mn containing melts intensively	
	–	–	The effect of Mn as an additive for hot-pressing process was studied *See also* Table 5.28 *See also* section C–Mn–Zr in Table I-2.14	

(continued)

Table 5.24 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	Reference
ZrC_{1-x}–Mo	–	1100	ZrC_{1-x} layer (thickness – 4 μm) on Mo wire is intact (compatible) after 150 h exposure	[1, 4, 30, 32, 33, 63, 158, 200, 213, 377, 870–872, 885, 901, 922, 1037–1043, 1051, 1058, 1230, 1243, 1429, 1460, 1463, 1523, 1541]
	–	1250	The max. solid solubility of ZrC_{1-x} in Mo is ~0.6 mol.%	
	–	1300	Under pressure 5 MPa a diffusion welded joint between ZrC_{1-x} and Mo parts is produced (5–15 min exposure)	
	Vacuum	1900–2000	The initiation of interaction between powdered (or sintered) carbide and compact dense metal with the formation of Mo solid solution in ZrC_{1-x} (2–5 h exposure)	
	–	1970	The max. solid solubility of ZrC_{1-x} in Mo is ~1.0 mol.%	
	–	2000–2200	Compact dense Mo and ZrC_{1-x} are compatible with each other; only weak diffusion interaction is observed (2–5 h exposure)	
	–	2200	No interaction between dispersed ZrC_{1-x} inclusions (up to 10%) with Mo matrix	
	–	–	Stable and compatible with each other as cermet components	
	–	>2200	The noticeable interaction of powdered carbide with compact dense metal is observed	
	–	~2240–2310	Eutectic ZrC_{1-x}–Mo; the max. solid solubility of Mo in ZrC_{1-x} is >10 mol.% and that of ZrC_{1-x} in Mo is ~3.2 mol.%	
	–	>2650	ZrC_{1-x} materials are corroded by liquid Mo and Mo containing melts *See also* section C–Mo–Zr in Table I-2.14	
ZrC_{1-x}–Mo–Ni	Vacuum, 0.1 Pa	1450	The contact interaction of compact dense ZrC_{1-x} (x = 0.04) with Ni – 25% Mo melt leads to the preferential dissolution of C; after cooling the needle-like isolations based on β-$Mo_{2\pm x}C$ (hexagonal) phase, surrounded by metal-semicarbide eutectics, are formed in the metal melt ($Ni_{\sim 0.91}Mo_{0.08}Zr_{0.004}C_{<0.005}$) in parallel with Ni–$ZrNi_{5\pm x}$ eutectics	[1525]
ZrC_{1-x}–Mo–Re	Vacuum	2450	No interaction in the contact zone (1 h exposure) between compact dense ZrC_{1-x} materials and Mo – 40% Re alloy	[993]
	He	2500	Interaction in the contact zone (15 min. exposure) between the compact dense materials	
ZrC_{1-x}–Mo–W		1200–2000	The solubility of ZrC_{1-x} in the Mo-W solid solution decreases with fall in temperature; homogeneity range of the solid solution steadily decreases with rise in W content	[943, 1429]
	–	–	Stable and compatible with each other as cermet components *See also* section C–Mo–W–Zr in Table I-2.14	

(continued)

Table 5.24 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	Reference
ZrC_{1-x}–TiC_{1-x}–Mo–Ni	Vacuum, 0.05 Pa	1450	The interaction of $Zr_{0.25}Ti_{0.75}C_{0.96}$ phase with Ni – 20–25% Mo melt is characterized by preferential dissolution of C and subsequent precipitation of $Zr_{0.01}Ti_{\sim0.60}Mo_{\sim0.40}C_{1-x}$ (from the melt saturated with Zr, Ti and C); the interaction of the compositions, having higher Zr content in them, with Ni-Mo melts such as $Zr_{0.50}Ti_{0.50}C_{0.96}$ and $Zr_{0.76}Ti_{0.24}C_{0.95}$, does not vary considerably, but after cooling in these cases the bulk content of $ZrNi_{5\pm x}$ intermetallide in solidified eutectics increases	[1066, 1525]
ZrC_{1-x}–TiC_{1-x}–δ-$TiN_{1\pm x}$–Mo–Ni	Vacuum, 0.01 Pa	1450	The interaction of complex carbonitrides $Zr_{0.06–0.20}Ti_{0.80–0.94}C_{0.50–0.80}N_{0.20–0.50}$ with Ni – 20% Mo melt leads to the accelerated incongruent dissolution of the components, such as C and Zr, preferentially; during the cooling procedure at the ceramic-metal interface, primary eutectic precipitates on the basis of β-$Mo_{2\pm x}C$ (hexagonal) and $(Ti,Mo)(C,N)_{1-x}$ (cubic) phases and $ZrNi_{5\pm x}$ and $TiNi_3$ intermetallides are formed (Zr is mainly concentrated in the metallic solid solutions and intermetallides)	[1524]
ZrC_{1-x}–$UC_{1\pm x}$–Mo	–	~2000	$(Zr_{0.50}U_{0.50})C_{1\pm x}$ reacts vigorously with metallic Mo *See also* section C–Mo–U–Zr in Table I-2.14	[928]
ZrC_{1-x}–Nb	Vacuum	900–1500	Interaction in the powdered mixtures (mean grain size – 1–2 μm) of components leads to the formation of β-$Nb_{2+x}C$	[1, 30, 32, 33, 42, 63, 213, 377, 901, 922, 1005, 1033–1036, 1067, 1150, 1230]
	–	1400	Under pressure 5 MPa a diffusion welded joint between ZrC_{1-x} and Nb is produced (5–15 min exposure)	
	Vacuum	1400–1600	The weak interaction of powdered carbide with compact dense metallic Nb is observed (5 h exposure); the formation of solid solutions of C in Nb in the contact zone between the dense compact components	
	–	1600–1800	The weak interaction of powdered carbide with compact dense metallic Nb results in the formation of ZrC_{1-x}–NbC_{1-x} solid solutions	
	–	2000–2200	The noticeable interaction with the formation of ZrC_{1-x}–NbC_{1-x} solid solutions; the max. solid solubility of Nb in ZrC_{1-x} is ~20–40 mol.%	
	Vacuum	2120	Eutectic ZrC_{1-x}–Nb	
	–	>2470	ZrC_{1-x} materials are corroded by liquid Nb and Nb containing melts *See also* Table 5.27 *See also* section C–Nb–Zr in Table I-2.13	

(continued)

Table 5.24 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	Reference
ZrC_{1-x}–$UC_{1\pm x}$–Nb	–	~2000	$(Zr_{0.50}U_{0.50})C_{1\pm x}$ reacts vigorously with metallic Nb *See also* section C–Nb–U–Zr in Table I-2.14	[928]
ZrC_{1-x}–Ni	–	700	The max. solid solubility of ZrC_{1-x} in Ni is ~0.6 mol.%	[1, 4, 30, 32, 33, 42, 45, 48, 63, 200, 850, 874–880, 901, 1002, 1003, 1025, 1026, 1063–1065, 1150, 1226, 1230, 1233, 1351, 1525]
	–	1100	Formation of a liquid phase in the contact zone between a carbide powder and compact dense Ni; the max. solid solubility of ZrC_{1-x} in Ni is ~1 mol.%	
	Ar	1160–1185	Formation of a liquid phase in the $ZrC_{\sim 1.0}$–Ni powdered mixture with mean grain size ~50 μm	
	–	~1210–1350	Eutectic ZrC_{1-x}–Ni; the max. solid solubility of ZrC_{1-x} in Ni is ~3.5 mol.% and that of Ni in ZrC_{1-x} is ~5 mol.%	
	–	1250	The max. solid solubility of ZrC_{1-x} in Ni is ~1.7 mol.%	
	Vacuum	~1280	Eutectic ZrC_{1-x}–Ni (C – 0.7–7.2%); the contact reaction leads to the penetration of eutectic alloy into the carbide substrate, the depth of penetration decreases with increasing C content in Ni–C alloy	
	–	1350–1400	The max. solid solubility of Ni in ZrC_{1-x} is ~3.5–4.3 mol.%; the addition of C (up to 7%) to Ni inhibits the penetration of Ni into the bulk ZrC_{1-x} and reduces the depth of the resultant diffusion zone in the contact, the interfacial reaction leads to the formation of Ni-based solid solutions, carbide-metal eutectics and intermetallide $ZrNi_{5\pm x}$ phase (the variation of Ni concentration between the boundary of a grain and its core obeys a parabolic law)	
	–	1450–1460	The contact interaction is very intensive (5 h exposure); the mutual exchange (dissolution) between the solid (ZrC_{1-x}) and liquid (Ni) phases was observed in the contact zone	
	N_2 + CO	1500	Intensive interaction with the dissolution of Ni and formation solid solutions (in the contact zone carbide has a changed lattice parameter) after 0.2 h exposure	
	–	>1500	ZrC_{1-x} materials are corroded by liquid Ni and Ni containing melts Data available in literature are controversial *See also* Table 5.28 *See also* section C–Ni–Zr in Table I-2.14	

(continued)

Table 5.24 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	Reference
ZrC_{1-x}–Ni–P			*See* Table 5.28	
ZrC_{1-x}–Ni–W			*See* Table 5.28	
ZrC_{1-x}–HfC_{1-x}–Ni			*See* section HfC_{1-x}–ZrC_{1-x}–Ni in Table 3.19	
			See also section C–Hf–Ni–Zr in Table I-2.14	
ZrC_{1-x}–β-$Mo_{2\pm x}C$–TiC_{1-x}–Ni	Vacuum	1250	The effect of ZrC_{1-x} addition on cermet sintering was studied	[1023]
ZrC_{1-x}–NbC_{1-x}–Ni			*See* section NbC_{1-x}–ZrC_{1-x}–Ni in Table 4.18	
			See also section C–Nb–Ni–Zr in Table I-2.14	
ZrC_{1-x}–TaC_{1-x}–Ni			*See* section TaC_{1-x}–ZrC_{1-x}–Ni in Table 2.21	
ZrC_{1-x}–TiC_{1-x}–Ni	–	~1160–1190	Eutectic $(Zr,Ti)C_{1-x}$–$(Ti,Zr)C_{1-x}$–Ni (cast alloys show a tendency to the crystallization in metastable state when the carbide phase does not decompose into two solid solutions)	[876, 880, 1063–1065, 1535]
	Ar	1230	Formation of a liquid phase in the powdered mixture $Zr_{0.25}Ti_{0.75}C_{1-x}$–Ni	
	Vacuum, 0.01–0.1 Pa	1450	$Zr_{0.25}Ti_{0.75}C_{\sim 1.0}$ phase has a higher dissolution rate in molten Ni; preferentially C and Zr are transfered into the melt, when attaining composition $Zr_{0.06}Ti_{0.94}C_{<0.96}$ the dissolution becomes congruent	
ZrC_{1-x}–TiC_{1-x}–δ-$TiN_{1\pm x}$–Ni	Vacuum, 0.01–0.1 Pa	1450	The interaction of complex carbonitrides $Zr_{0.06–0.20}Ti_{0.80–0.94}C_{0.50–0.80}N_{0.20–0.50}$ with molten Ni leads to the incongruent dissolution of the components, preferentially – C and Zr; so the outer layers of complex carbonitride grains are enriched with N and Ti – up to $Zr_{0.01–0.02}Ti_{0.98–0.99}C_{0.05–0.15}N_{0.85–0.95}$ and delaminated, the metallic melts have a hypoeutectic structure with primary precipitates of Ni based phase and ternary eutectics (no intermetallide phases were observed)	[1435, 1524]
ZrC_{1-x}–TiC_{1-x}–δ-$TiN_{1\pm x}$–δ-$WC_{1\pm x}$–Ni	Vacuum	1450	The formation of complex monocarbonitride $(Zr,Ti,W)(C,N)_{1-x}$ phases with core-rim microstructures	[945]
ZrC_{1-x}–TiC_{1-x}–δ-$TiN_{1\pm x}$–$TiO_{1\pm x}$–Ni	Vacuum, 10^{-1} Pa	1450	The interaction of $Zr_{0.05}Ti_{0.95}C_{0.50}N_{0.40}O_{0.10}$ oxycarbonitride phase with Ni is considered	[1435]
ZrC_{1-x}–δ-$WC_{1\pm x}$–Ni	–	1000	General consideration of the system; regions of solid solutions based on the individual components are not large (~6–8 mol.%)	[875]
			See also section C–Ni–W–Zr in Table I-2.14	

(continued)

Table 5.24 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	Reference
ZrC_{1-x}–$UC_{1\pm x}$–Ni	–	–	The effect of Ni as an additive for hot-pressing process was studied	[1357]
ZrC_{1-x}–$ZrB_{2\pm x}$–Ni	–	–	Plasma-sprayed coatings were fabricated and studied; the positive effect of Ni on the densification during reactive hot-pressing process was observed	[886, 1416, 1458, 1459]
ZrC_{1-x}–Os	–	1500	ZrC_{1-x} is in equilibrium with metal Os; the mutual solid solubilities of ZrC_{1-x} and Os are low, η-$Zr_4Os_2C_{1-x}$ is formed	[200, 990, 991, 1531]
	–	–	ZrC_{1-x} is decomposed to form graphite and solid solution of Zr in metallic Os *See also* section C–Os–Zr in Table I-2.14	
ZrC_{1-x}–Pb	Ar	≤400	No chemical interaction (10 h exposure)	[1, 30, 32, 33, 42, 45, 48, 63, 836–839, 850]
	Air	450	Traces of Zr were revealed in liquid metal after 10 h exposure	
	–	–	Formation of Zr_2PbC ($M_{n+1}AX_n$-phase) *See also* Table 5.28 *See also* section C–Pb–Zr in Table I-2.14	
ZrC_{1-x}–Pd	–	≥1300	The interaction with Pd results in the formation of ordered $ZrPd_3$ intermetallide phase and C (graphite) precipitation *See also* Table 5.27 *See also* section C–Pd–Zr in Table I-2.14	[200, 1014]
ZrC_{1-x}–Pt	–	≥1500	The interaction with Pt results in the formation of ordered $ZrPt_{3\pm x}$ intermetallide phase and C (graphite) precipitation *See also* section C–Pt–Zr in Table I-2.14	[200, 990, 991]
ZrC_{1-x}–Pu			*See* section C–Pu–Zr in Table I-2.14 ZrC_{1-x}–$UC_{1\pm x}$–Pu *See* section C–Pu–U–Zr in Table I-2.14	
ZrC_{1-x}–Rb	–	–	ZrC_{1-x} materials are stable in liquid and vapourized Rb in the wide temperature range	[63]
ZrC_{1-x}–Re	–	1900	The solubility of Re in carbide is ~3 mol.% and that of ZrC_{1-x} in Re is <0.5 mol.%	[1, 200, 872, 881, 882, 901, 993, 1033, 1047, 1429]
	Vacuum	2500	No interaction in the contact zone between the compact dense materials (1 h exposure)	
	–	–	Stable and compatible with each other as cermet components	
	–	~2540–2670	Eutectic ZrC_{1-x}–Re; the max. solid solubility of Re in ZrC_{1-x} is 1.7 mol.% Data available in literature are controversial *See also* section C–Re–Zr in Table I-2.14	
ZrC_{1-x}–$UC_{1\pm x}$–Re	–	2100	$(Zr_{0.5}U_{0.5})C_{1\pm x}$ is compatible with metallic Re (up to ~0.5 h exposure) *See also* section C–Re–U–Zr in Table I-2.14	[928]

(continued)

Table 5.24 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	Reference
ZrC_{1-x}–Rh	–	≥1500	The interaction results in the formation of $ZrRh_{3\pm x}$ (or $ZrRh_3C_x$) and C (graphite); practically, no solubility of Rh in ZrC_{1-x} *See also* section C–Rh–Zr in Table I-2.14	[200]
ZrC_{1-x}–Ru	–	≥1500	The interaction of non-stoichiometric ZrC_{1-x} results in the formation of $ZrRu_3C_{1-x}$ ($x \approx 0.3$); quasi-stoichiometric phases are in equilibrium with Ru	[200]
	–	~1800	Eutectic ZrC_{1-x}–Ru *See also* Table 5.27 *See also* section C–Ru–Zr in Table I-2.14	
ZrC_{1-x}–S	–	–	Formation of Zr_2SC ($M_{n+1}AX_n$-phase) *See also* section C–S–Zr in Table I-2.14	[836–838]
ZrC_{1-x}–Sb	Ar	<700	No chemical interaction *See also* Table 5.28	[1, 850]
ZrC_{1-x}–Sc			*See* section C–Sc–Zr in Table I-2.14	
ZrC_{1-x}–Si	Vacuum	300–800	Nanocrystalline ZrC_{1-x} films (mean grain size < 5 nm) thermally stable (0.5 h exposure) in contact with Si substrate	[1, 30, 32, 33, 42, 45, 48, 63, 76, 883, 884, 902, 995, 1025, 1026, 1060, 1083, 1233, 1245]
	Ar	800–1000	Weak interaction in the powdered mixtures of components with the formation of $ZrSi_2$ (1 h exposure, growing from 1.2 to 12.1% with increasing temperature)	
	Ar	1100	The interaction in the powdered mixtures results in the formation of α-ZrSi (17%) and small amounts of β-SiC (1%)	
	Ar	1400–1700	The intensive interaction in the powdered mixtures results in the formation of α/β-ZrSi and β-SiC (0.2–1.0 h exposure, no non-combined Si in the products)	
		1410	The mutual exchange (dissolution) between the solid (ZrC_{1-x}) and liquid (Si) phases was observed in the contact zone	
	–	>1450–1500	ZrC_{1-x} materials are corroded by liquid Si and Si containing melts intensively *See also* Table 5.28 *See also* section C–Si–Zr in Table I-2.14	
ZrC_{1-x}–Al_4C_3–Si	–	–	Formation of $Zr(Al_{1-x}Si_x)_8C_7$, $Zr(Al_{1-x}Si_x)_4C_4$, $Zr_2(Al_{1-x}Si_x)_4C_5$ and $Zr_3(Al_{1-x}Si_x)_4C_6$ *See also* section ZrC_{1-x}–SiC–Al in this table *See also* section ZrC_{1-x}–Al_4C_3–SiC in Table 5.25	[826, 827, 831–833, 904, 1072, 1082]

(continued)

Table 5.24 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	Reference
ZrC_{1-x}–Sn	Air	≤350	No chemical interaction (10 h exposure)	[1, 30, 32, 33, 45, 48, 63, 836–839, 850, 1529]
	–	–	Formation of Zr_2SnC ($M_{n+1}AX_n$-phase) and Zr_5Sn_3C (?)	
			See also Table 5.28	
			See also section C–Sn–Zr in Table I-2.14	
ZrC_{1-x}–Sr			*See* Table 5.27	
ZrC_{1-x}–Ta	–	1600–2000	No interaction in the contact zone between the compact dense materials (2 h exposure)	[1, 32, 33, 42, 63, 76, 200, 213, 886–888, 901, 922, 995, 1044, 1045, 1081, 1230]
	–	1900	Under pressure 5 MPa a diffusion welded joint between ZrC_{1-x} and Ta is produced (5–15 min exposure)	
	–	2000–2200	Weak interaction in the contact zone between the compact dense materials (2–5 h exposure)	
	Vacuum	>2200	The interaction of powdered carbide with compact dense metal results in the formation of ZrC_{1-x}–TaC_{1-x} solid solutions	
	–	>3000	ZrC_{1-x} materials are corroded by liquid Ta and Ta containing melts	
			See also section C–Ta–Zr in Table I-2.14	
ZrC_{1-x}–$UC_{1\pm x}$–Ta	–	~2100	$(Zr_{0.50}U_{0.50})C_{1\pm x}$ is compatible with metallic Ta for periods of up to 35 min	[928]
			See also section C–Ta–U–Zr in Table I-2.14	
ZrC_{1-x}–Tc			*See* section C–Tc–Zr in Table I-2.14	
ZrC_{1-x}–α/β-Th	–	1500	ZrC_{1-x} is in equilibrium with both α-Th and β-Th metal modifications; the mutual solid solubilities of the components are low	[200, 950–952, 1006]
			See also section C–Th–Zr in Table I-2.14	
ZrC_{1-x}–α/β-Ti	–	1730	The contact interaction is very intensive (5 h exposure)	[48, 200, 888–893]
	Vacuum, $< 7 \times 10^{-4}$ Pa	1850	ZrC_{1-x} materials are highly soluble in the melt Ti–Zr–C alloys formed at the interface (powerful solvent action of liquid Ti)	
			See also section C–Ti–Zr in Table I-2.14	
ZrC_{1-x}–Tl	Ar	<400	No chemical interaction	[1, 836–839, 850]
	–	–	Formation of Zr_2TlC ($M_{n+1}AX_n$-phase)	
			See also Table 5.28	
			See also section C–Tl–Zr in Table I-2.14	
ZrC_{1-x}–α/β/γ-U	–	>1150	ZrC_{1-x} materials are corroded by liquid U and U containing melts	[63]
			See also section C–U–Zr in Table I-2.14	

(continued)

Table 5.24 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	Reference
ZrC_{1-x}–V	–	1685	Eutectic (Zr,V)C_{1-x}–(V,Zr)	[4, 48, 63, 1429, 1523]
	–	1730	Contact interaction is very intensive (5 h exposure)	
	–	>1900	ZrC_{1-x} materials are corroded by liquid V and V containing melts	
	–	–	Stable and compatible with each other as cermet components	
			See also section C–V–Zr in Table I-2.14	
ZrC_{1-x}–W	–	1600–2000	No interaction between powdered ZrC_{1-x} and compact dense metallic W	[1, 4, 32, 33, 200, 213, 377, 576, 871, 894–899, 901, 922, 993, 1041, 1046, 1230, 1243, 1362, 1429, 1523, 1549]
	–	1800	Under pressure 5 MPa a diffusion welded joint between ZrC_{1-x} and W parts is produced (5–15 min exposure)	
	Vacuum	2200	The interaction between powdered (or sintered) ZrC_{1-x} and compact dense metal leads to the formation of W-C solid solutions (2–5 h exposure)	
	–	–	Stable and compatible with each other as cermet components	
	He	2500	No interaction in the contact zone between the compact dense materials (2 h exposure)	
	–	~2630–2800	Eutectic ZrC_{1-x}–W; the max. solubility of W in ZrC_{1-x} is ~7 mol.%	
			See also section C–W–Zr in Table I-2.14	
ZrC_{1-x}–$UC_{1\pm x}$–W	–	~2100	$(Zr_{0.50}U_{0.50})C_{1\pm x}$ is compatible with metallic W for periods of up to 35 min	[928]
			See also section C–U–W–Zr in Table I-2.14	
ZrC_{1-x}–γ-$W_{2\pm x}C$–W	–	2620	Eutectic ZrC_{1-x}–γ-$W_{2\pm x}C$–W	[894, 896, 899]
			See also section C–W–Zr in Table I-2.14	
ZrC_{1-x}–Zn	Ar	550	The interaction is weak: after 6 h exposure the content of Zr in the melt achieved 0.02%	[30, 32, 33, 42, 45, 48, 63]
	–	>550	ZrC_{1-x} materials are corroded by liquid Zn and Zn containing melts slightly	
		940	Weak contact interaction (170 h exposure)	
ZrC_{1-x}–α/β-Zr	–	1300–1700	The formation of solid solutions of C in Zr in the contact zone (the microhardness of carbide decreases and that of metal increases)	[1, 10, 48, 126, 600, 900, 901, 1005, 1057, 1085, 1370, 1371, 1375]
	–	1805–1835	Eutectic ZrC_{1-x}–β-Zr with actual compositions $ZrC_{0.49-0.57}$–$Zr(C_{\sim 0.02})$	
	–	1950	The contact interaction is very intensive (5 h exposure)	
	–	~2000	The molten metallic Zr caused a very limited dissolution of ZrC_{1-x}	
			See also section C–Zr in Table I-2.13	

(continued)

Table 5.24 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	Reference
ZrC_{1-x}–$ZrN_{1\pm x}$–α/β-Zr	–	1150	Monocarbonitride phases with ~$ZrC_{0.56}$ – ~$ZrC_{0.33}N_{0.23}$ compositions are in the equilibrium with metallic solid solutions ranged from β-Zr($C_{\sim 0.01}$) to β-Zr($C_{\sim 0.01}$,$N_{\sim 0.02}$), while monocarbonitride phases with ~$ZrC_{0.33}N_{0.23}$ – ~$ZrN_{0.56}$ compositions are in the equilibrium with metallic solid solutions ranged from α-Zr($C_{\sim 0.02}$,$N_{\sim 0.14}$) to α-Zr($N_{\sim 0.30}$) *See also* section C–N–Zr in Table I-2.14	[126, 962, 994]
ZrC_{1-x}–$UC_{1\pm x}$–U–Zr	–	1700	The composition of (Zr,U)$C_{1\pm x}$ in the equilibrium with Zr–U alloy varies with increasing U content rapidly from $ZrC_{0.62}$ to (Zr,U)$C_{0.88}$ and remains constant up to $UC_{0.88}$ *See* section C–U–Zr in Table I-2.14	[958]
ZrC_{1-x}–$ZrB_{2\pm x}$–α/β-Zr	–	~1615–1650	Eutectic ZrC_{1-x}–$ZrB_{2\pm x}$–β-Zr *See also* section C–B–Zr in Table I-2.14	[200, 840, 843, 845, 1086]

[a]The parameters of wettability of ZrC_{1-x} phases by liquid metals at various temperatures are listed in Table 5.28

Table 5.25 Chemical interaction and/or compatibility of zirconium carbide phases with refractory compounds at elevated, high and ultra-high temperatures (reaction systems are given mainly in alphabetical order)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	Reference
ZrC_{1-x}–Al_4C_3	–	up to 2200	No mutual solid solubilities between the components	[814–835, 904–913, 1011]
	–	–	Formation of $Zr_3Al_3C_5$, $Zr_2Al_3C_5$, Zr_5Al_3C, $Zr_2Al_3C_4$, $ZrAl_8C_7$, $ZrAl_4C_4$ and $ZrAlC_{2-x}$ *See also* section ZrC_{1-x}–Al in Table 5.24 *See also* section C–Al–Zr in Table I-2.14	
ZrC_{1-x}–Al_4C_3–SiC	–	–	Formation of intergrowth structures, consisting of Zr_mC_{m+1} (NaCl-type) slabs separated by $(Al_{1-z}Si_z)_8C_7$ (Al_4C_3-type) layers, such as $Zr(Al_{1-x}Si_x)_8C_7$ ($0 \le x \le 0.07$), $Zr(Al_{1-x}Si_x)_4C_4$ ($0 \le x \le 0.04$), $Zr_2(Al_{1-x}Si_x)_4C_5$ ($0 \le x \le 0.11$) and $Zr_3(Al_{1-x}Si_x)_4C_6$ ($0 \le x \le 0.11$) *See also* section ZrC_{1-x}–SiC–Al in Table 5.24 *See also* section ZrC_{1-x}–Al_4C_3–Si in Table 5.24	[826, 827, 831–833, 904, 910, 1082]
ZrC_{1-x}–Al_4C_3–SiC–TiC_{1-x}	–	–	Formation of $(Ti,Zr)_3(Si,Al)C_2$ ($M_{n+1}AX_n$-phase solid solution) *See also* section ZrC_{1-x}–SiC–TiC_{1-x}–Al in Table 5.24	[834, 838]

(continued)

Table 5.25 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	Reference
ZrC_{1-x}–Al_4C_3–$YC_{1\pm x}$/β-$YC_{2\pm x}$	–	–	Formation of $(Zr,Y)Al_4C_4$ and $(Zr,Y)_2Al_4C_5$	[908, 909]
ZrC_{1-x}–α-Al_2O_3	CO_2, 1.4×10^{-17} Pa	1000	Calculated equilibrium pressure of the interaction between the components	[381, 903, 1015]
	–	1530	Interaction between the components during a hot-pressing procedure was studied	
ZrC_{1-x}–$B_{4\pm x}C$	–	1400–1500	Formation of $ZrB_{2\pm x}$ and C (via a stage of intermediate phase formation)	[33, 42, 200, 840, 841, 843–845, 1030, 1462]
	Chamber of anvil type	1600–2000 (15 MPa)	Formation of $ZrB_{2\pm x}$	
		1750–1950 (5 GPa)	Formation of $ZrB_{2\pm x}$; as the temperature increases, the ratio between the ZrC (remain) and ZrB_2 (product) phases alters in a such way that the ZrB_2 content increases *See also* section C–B–Zr in Table I-2.14	
ZrC_{1-x}–$B_{4\pm x}C$–α-BN	–	≤ 2100	No contact reaction between powdered ZrC_{1-x} and compact dense $B_{4\pm x}C$–α-BN composite	[915, 916, 996]
ZrC_{1-x}–$B_{4\pm x}C$–$SiB_{3\pm x}$	Chamber of anvil type	1600–2000 (0.015–5 GPa)	Formation of $ZrB_{2\pm x}$; the addition of $SiB_{3\pm x}$ favours the reaction between ZrC_{1-x} and $B_{4\pm x}C$ going to its completion *See also* section C–B–Si–Zr in Table I-2.14	[1030]
ZrC_{1-x}–α-BN	–	≥ 2400	Formation of $ZrB_{2\pm x}$	[33, 914–916]
ZrC_{1-x}–CaO	Vacuum	1245	The rate of the following reaction (confirmed experimentally for $x = 0.04$ and 0.09) $ZrC_{1-x} + (4 - x)CaO = CaZrO_3 + (3–x)Ca\uparrow +$ $+ (1 - x)CO\uparrow$ is controlled by diffusion of reactant ions through the product layer of $CaZrO_3$	[997]
ZrC_{1-x}–β-$CeC_{2\pm x}$	–	1600	No mutual solubilities *See also* section C–Ce–Zr in Table I-2.14	[200]
ZrC_{1-x}–β-$CeC_{2\pm x}$–$UC_{1\pm x}$	–	–	The solubility of "imaginary" phase 'CeC' in $(U,Zr,Ce)C_{1\pm x}$ mixed carbide decreases sharply with increasing content of Zr *See also* section C–Ce–U–Zr in Table I-2.14	[200]
ZrC_{1-x}–$CeN_{1\pm x}$	–	–	Terminal mutual solid solubilities between the components, ?	[33]
ZrC_{1-x}–$CeP_{1\pm x}$	–	–	Terminal mutual solid solubilities between the components, ?	[33]
ZrC_{1-x}–$CeS_{1\pm x}$	–	–	Terminal mutual solid solubilities between the components, ?	[33]

(continued)

Table 5.25 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	Reference
ZrC_{1-x}–Cr_3C_{2-x}	–	1300–1800	The max. solubility of "imaginary" phase 'CrC_{1-x}' in ZrC_{1-x} is ~1.5–3.0 mol.% and that of ZrC_{1-x} in Cr_3C_{2-x} is very low; ZrC_{1-x} phase is in equilibrium with C (graphite) and Cr_3C_{2-x} *See also* section C–Cr–Zr in Table I-2.14	[33, 200, 552, 856, 860–863, 995]
ZrC_{1-x}–$Cr_7C_{3\pm x}$	–	1750	Eutectic ZrC_{1-x}–$Cr_7C_{3\pm x}$; ZrC_{1-x} phase is in equilibrium with all Cr carbides, intermetallide and metallic Cr, the max. solubility of Cr in ZrC_{1-x} is ~1.5–3.0 mol.% and that of Zr in $Cr_7C_{3\pm x}$ is very low *See also* section C–Cr–Zr in Table I-2.14	[200, 856, 860–863, 917]
ZrC_{1-x}–Cr_3C_{2-x} –$Cr_7C_{3\pm x}$	–	1730	Eutectic ZrC_{1-x}–Cr_3C_{2-x}–$Cr_7C_{3\pm x}$ *See also* section C–Cr–Zr in Table I-2.14	[860–863, 917]
ZrC_{1-x}–$Cr_{23}C_{6\pm x}$	–	–	ZrC_{1-x} phase is in equilibrium with all Cr carbides and intermetallide and metallic Cr, the max. solid solubility of Cr in ZrC_{1-x} is ~1.5–3.0 mol.% and that of Zr in $Cr_{23}C_{6\pm x}$ is very low *See also* section ZrC_{1-x}–$Cr_{23}C_{6\pm x}$–Cr in Table 5.24 *See also* section C–Cr–Zr in Table I-2.14	[200, 860–863, 917–919]
ZrC_{1-x}–Cr_3C_{2-x} –TiC_{1-x}		1550	The addition of ZrC_{1-x} into Cr_3C_{2-x} matrix suppresses the recrystallization due to the formation of $(Ti,Cr)C_{1-x}$ in the presence of TiC_{1-x}, because of the difference in lattice parameters between TiC_{1-x} and the mixed (solid solution) monocarbide is diminished by forming $(Ti,Cr,Zr)C_{1-x}$ carbide phase	[1538]
ZrC_{1-x}–Cr_2O_3	CO_2, 6.5×10^{-9} Pa	1000	Calculated equilibrium pressure of the interaction between the components	[381, 903]
ZrC_{1-x}–$DyN_{1\pm x}$	–	–	Monocarbonitride (cubic) continuous solid solution (complete solubility in the system?)	[33]
ZrC_{1-x}–Dy_2O_3	Vacuum, ~3 Pa	1500–1700	The effect of Dy_2O_3 as an additive for spark plasma sintering process was studied	[1394]
ZrC_{1-x}–$ErN_{1\pm x}$	–	–	Monocarbonitride (cubic) continuous solid solution (complete solubility in the system?)	[33]
ZrC_{1-x}–$EuN_{1\pm x}$	–	–	Monocarbonitride (cubic) continuous solid solution (complete solubility in the system?)	[33]
ZrC_{1-x}–EuO	–	–	Terminal mutual solid solubilities between the components (?)	[33]
ZrC_{1-x}–Fe_3C	–	–	Not (or very slightly) soluble in each other in solid state *See also* section C–Fe–Zr in Table I-2.14	[43, 200]

(continued)

Table 5.25 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	Reference
ZrC_{1-x}–Fe_2O_3	Vacuum	800–900	Interaction leads to the formation of C-depleted ZrC_{1-x} phases, α-ZrO_2, Fe_3O_4 and α-Fe	[381, 903, 925]
	CO_2, 6.8×10^{-3} Pa	1000	Calculated equilibrium pressure of the interaction between the components	
ZrC_{1-x}–$GdN_{1\pm x}$	–	–	Monocarbonitride (cubic) continuous solid solution (complete solubility in the system?)	[33]
ZrC_{1-x}–HfC_{1-x}			*See* section HfC_{1-x}–ZrC_{1-x} in Table 3.20	
ZrC_{1-x}–HfC_{1-x}–NbC_{1-x}			*See* section HfC_{1-x}–NbC_{1-x}–ZrC_{1-x} in Table 3.20	
ZrC_{1-x}–HfC_{1-x}–TaC_{1-x}			*See* section TaC_{1-x}–HfC_{1-x}–ZrC_{1-x} in Table 2.22	
ZrC_{1-x}–HfC_{1-x}–$ThC_{1\pm x}$			*See* section HfC_{1-x}–$ThC_{1\pm x}$–ZrC_{1-x} in Table 3.20	
ZrC_{1-x}–HfC_{1-x}–TiC_{1-x}			*See* section HfC_{1-x}–TiC_{1-x}–ZrC_{1-x} in Table 3.20	
ZrC_{1-x}–HfC_{1-x}–$UC_{1\pm x}$			*See* section HfC_{1-x}–$UC_{1\pm x}$–ZrC_{1-x} in Table 3.20 *See also* section C–Hf–U–Zr in Table I-2.14	
ZrC_{1-x}–HfC_{1-x}–VC_{1-x}			*See* section HfC_{1-x}–VC_{1-x}–ZrC_{1-x} in Table 3.20	
ZrC_{1-x}–HfC_{1-x}–δ-$WC_{1\pm x}$			*See* section HfC_{1-x}–δ-$WC_{1\pm x}$–ZrC_{1-x} in Table 3.20	
ZrC_{1-x}–$HfN_{1\pm x}$	–	1600–2200	Monocarbonitride (cubic) continuous solid solution (complete solubility in the system)	[33, 147]
ZrC_{1-x}–$HoN_{1\pm x}$	–	–	Monocarbonitride (cubic) continuous solid solution (complete solubility in the system?)	[33]
ZrC_{1-x}–$LaN_{1\pm x}$	–	–	Terminal mutual solid solubilities between the components (?)	[33]
ZrC_{1-x}–$LaB_{6\pm x}$	Vacuum	1900	Interaction between the powders (exposure – 0.5 h) leads to the formation of $ZrB_{2\pm x}$ and some $(La,Zr)_x(C,B)_y$ unidentified phases; no changes in ZrC_{1-x} stoichiometry	[1457]
ZrC_{1-x}–$LaB_{6\pm x}$–SiC	Vacuum	1900	Interaction between the powders ZrC_{1-x} and $LaB_{6\pm x}$ (exposure – 0.5 h) leads to the formation of $ZrB_{2\pm x}$; SiC powder is almost inert to other ingredients of the composition	[1457]
ZrC_{1-x}–$LaP_{1\pm x}$	–	–	Terminal mutual solid solubilities between the components (?)	[33]
ZrC_{1-x}–LaS_{1-x}	–	–	Terminal mutual solid solubilities between the components (?)	[33]
ZrC_{1-x}–$LuN_{1\pm x}$	–	–	Monocarbonitride (cubic) continuous solid solution (complete solubility in the system?)	[33]
ZrC_{1-x}–MgO	Vacuum	2000–2300	Interaction of compact dense materials results in the formation of ZrC_xO_y oxycarbide phases and MgC (?) in the contact zone; also other products (metallic Mg, graphite C) were observed *See also* section C–Mg–O–Zr in Table I-2.14	[1, 33, 923–925]

(continued)

Table 5.25 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	Reference
ZrC_{1-x}–Mn_5C_2, Mn_7C_3	–	700–1000	Not (or very slightly) soluble in each other at solid state *See also* section C–Mn–Zr in Table I-2.14	[43]
ZrC_{1-x}–α-MoC_{1-x}	–	1900–2000	Extended monocarbide (cubic) solid solution based on ZrC_{1-x}	[33, 117, 158, 200, 329, 870, 871, 882, 926, 927, 930]
	–	~2000–2600	Monocarbide (cubic) continuous solid solution	
			See also section C–Mo–Zr in Table I-2.14	
ZrC_{1-x}–α-MoC_{1-x}–NbC_{1-x}			*See* section NbC_{1-x}–α-MoC_{1-x}–ZrC_{1-x} in Table 4.19	
			See also section C–Mo–Nb–Zr in Table I-2.14	
ZrC_{1-x} – α-MoC_{1-x}–$UC_{1\pm x}$	–	2000	Extended monocarbide (cubic) solid solution based on ZrC_{1-x}–$UC_{1\pm x}$ continuous solid solution; the solubility of α-MoC_{1-x} in ZrC_{1-x} is >80 mol.% (the solubilities of ZrC_{1-x} and $UC_{1\pm x}$ in α-MoC_{1-x} are low) *See also* section C–Mo–U–Zr in Table I-2.14	[927]
ZrC_{1-x}–β-$Mo_{2\pm x}C$	–	1400–2100	Solubility of Zr in β-$Mo_{2\pm x}C$ is very low *See also* section C–Mo–Zr in Table I-2.14	[33, 158, 200, 870, 871, 930]
ZrC_{1-x}–$MoSi_2$	Ar	1900–1950	Interaction in powder mixtures leads to the formation of $(Mo,Zr)_5Si_{3\pm x}$ and SiC *See also* section C–Mo–Si–Zr in Table I-2.14	[476, 587, 588, 1181–1183, 1410, 1411]
ZrC_{1-x}–NbC_{1-x}			*See* section NbC_{1-x}–ZrC_{1-x} in Table 4.19	
ZrC_{1-x}–NbC_{1-x} –TaC_{1-x}			*See* section TaC_{1-x}–NbC_{1-x}–ZrC_{1-x} in Table 2.22	
ZrC_{1-x}–NbC_{1-x} –$ThC_{1\pm x}$			*See* section NbC_{1-x}–$ThC_{1\pm x}$–ZrC_{1-x} in Table 4.19	
ZrC_{1-x}–NbC_{1-x} –TiC_{1-x}			*See* section NbC_{1-x}–TiC_{1-x}–ZrC_{1-x} in Table 4.19	
ZrC_{1-x}–NbC_{1-x} –$UC_{1\pm x}$			*See* section NbC_{1-x}–$UC_{1\pm x}$–ZrC_{1-x} in Table 4.19 *See also* section C–Nb–U–Zr in Table I-2.14	
ZrC_{1-x}–NbC_{1-x} –VC_{1-x}			*See* section NbC_{1-x}–VC_{1-x}–ZrC_{1-x} in Table 4.19	

(continued)

Table 5.25 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	Reference
ZrC_{1-x}–NbC_{1-x} –δ-$WC_{1\pm x}$			*See* section NbC_{1-x}–δ-$WC_{1\pm x}$–ZrC_{1-x} in Table 4.19	
ZrC_{1-x}– β-$Nb_{2+x}C$			*See* section β-$Nb_{2+x}C$–ZrC_{1-x} in Table 4.19	
			See also section C–Nb–Zr in Table I-2.14	
ZrC_{1-x}– δ-NbN_{1-x}	N_2, >0.1 MPa	1500	Monocarbonitride (cubic) continuous solid solution (complete solubility in the system)	[33, 140, 1532]
ZrC_{1-x}–NbO	–	–	Terminal mutual solid solubilities between the components (?)	[33]
			See also section C–Nb–O–Zr in Table I-2.14	
ZrC_{1-x}–$NdN_{1\pm x}$	–	–	Monocarbonitride (cubic) continuous solid solution (complete solubility in the system?)	[33]
ZrC_{1-x}– $Nd_2Fe_{14}B$	Pure Ar	1000	The solubility of ZrC_{1-x} in complex boride phase is ~6 mol.%	[1055]
ZrC_{1-x}–NiO	–	–	Interaction between the components is determined by the carbide composition: $ZrC_{0.97}$ primarily reacts to form C oxides, then as the carbide phase is depleted of C, γ-ZrO_{2-x} is formed, while C is precipitated in non-combined form; but $ZrC_{0.77}$ reacts only through the preferential formation of Zr oxide (no oxycarbide phases were revealed in these conditions)	[1016]
ZrC_{1-x}–NpC_{1-x}	–	–	Monocarbide (cubic) continuous solid solution (complete solubility in the system?)	[33]
ZrC_{1-x}–$NpN_{1\pm x}$	–	–	Monocarbonitride (cubic) continuous solid solution (complete solubility in the system?)	[33]
ZrC_{1-x}–$PrN_{1\pm x}$	–	–	Monocarbonitride (cubic) continuous solid solution (complete solubility in the system?)	[33]
ZrC_{1-x}–PuC_{1-x}	–	1250–1500	Max. terminal solid solubility of ZrC_{1-x} in PuC_{1-x} is 23 mol.% and that of PuC_{1-x} in ZrC_{1-x} is 24 mol.%	[33, 116, 200, 936–939]
	–	>1600	Monocarbide (cubic) continuous solid solution (complete solubility in the system)	
			See also section C–Pu–Zr in Table I-2.14	
ZrC_{1-x}–PuC_{1-x}– $UC_{1\pm x}$	–	–	Monocarbide (cubic) continuous solid solution (in the regions with lower PuC_{1-x} content)	[200, 1019, 1355, 1356]
			See also section C–Pu–U–Zr in Table I-2.14	
ZrC_{1-x}–Pu_2C_{3-x}	–	1250–1600	Terminal solid solubility of ZrC_{1-x} is low	[200, 936–939]
			See also section C–Pu–Zr in Table I-2.14	
ZrC_{1-x}–$PuN_{1\pm x}$	–	–	Monocarbonitride (cubic) continuous solid solution (complete solubility in the system)	[33]

(continued)

Table 5.25 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	Reference
ZrC_{1-x}–$PuP_{1\pm x}$	–	–	Terminal mutual solid solubilities between the components (?)	[33]
ZrC_{1-x}–$PuS_{1\pm x}$	–	–	Terminal mutual solid solubilities between the components (?)	[33]
ZrC_{1-x}–ScC_{1-x}	–	1500–1900	Monocarbide (cubic) continuous solid solution (complete solubility in the system) *See also* section C–Sc–Zr in Table I-2.14	[33, 940, 941]
ZrC_{1-x}–$ScN_{1\pm x}$	–	–	Monocarbonitride (cubic) continuous solid solution (complete solubility in the system?)	[33]
ZrC_{1-x}–α/β-SiC	Vacuum, 10^{-3} Pa	1300	A transition zone formed on the ZrC_{1-x} side between compact dense components during diffusion welding procedure (loading time – 20 min), while annealing procedure (50 h exposure) the zone stratified forming a solid solution of Si in ZrC_{1-x} with SiC inclusions	[33, 76, 883, 884, 929, 935, 942, 1050, 1059, 1364, 1365, 1535]
	–	–	Practically, the pure components are compatible with each other within the temperature range of their thermal stability *See also* section C–Si–Zr in Table I-2.14	
ZrC_{1-x}–α/β-SiC –δ-$WC_{1\pm x}$	–	1800	Hot-pressing of powdered δ-$WC_{1.01}$ (mean particle size – 0.75 μm, contents: non-combined C – 0.04%, Fe – 0.05%) – 1–6 mol.% β-SiC (mean particle size – 0.4–0.6 μm, contents: O – 1.80%, Fe – 0.05%) – 2–14 mol.% $ZrC_{0.95}$ (mean particle size – 2 μm, contents: non-combined C – 0.25%, O – 0.62%, N – 0.68%, Fe – 0.01%) mixtures led to the formation of $(Zr,W)C_{1-x}$ solid solution and small amounts of $W_{2\pm x}C$ and ZrO_{2-x}	[1550]
ZrC_{1-x}–SiC–$ZrB_{2\pm x}$	–	1800–2000	The components are compatible to each other as constituents of composites	[1050, 1068, 1427, 1542, 1543]
ZrC_{1-x}–α/β-Si_3N_4	–	~1400–1600	Interaction with the formation of Zr silicides, β-SiC and monocarbonitride (cubic) phase with the composition close to ~$ZrC_{0.4}N_{0.6}$; carbide phase inhibits α → β nitride transformation	[1013, 1045, 1053, 1054, 1472]
ZrC_{1-x}–β-Si_3N_4–Al_2O_3	–	1650–1700	Interaction with the formation of β-SiC, monocarbonitride (cubic) phase with the composition close to ~$ZrC_{0.4}N_{0.6}$, Si_3N_4–Al_2O_3 solid solutions and glasses of $Al_2O_3{\cdot}SiO_2$ type	[1053, 1054]
ZrC_{1-x}–β-Si_3N_4–MgO	–	1650–1700	Interaction with the formation of β-SiC, monocarbonitride (cubic) phase with the composition close to ~$ZrC_{0.4}N_{0.6}$, $MgSiO_4$, Mg_2SiO_3 and glasses of $MgO{\cdot}SiO_2$ type	[1053, 1054]
ZrC_{1-x}–SiO_2–Al_2O_3–CaO–Fe_2O_3 (basalt)	–	1400	ZrC_{1-x} (8–12% porosity) is not stable in molten basalt (SiO_2 – 15% Al_2O_3 – 13% CaO – 13% Fe_2O_3) at short-term exposure (0.5–1 h) with average dissolution rate ~0.2 g cm^{-2} h^{-1}	[1475]

(continued)

Table 5.25 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	Reference
ZrC_{1-x}–$SmN_{1\pm x}$	–	–	Monocarbonitride (cubic) continuous solid solution (complete solubility in the system?)	[33]
ZrC_{1-x}–TaC_{1-x}			*See* section TaC_{1-x}–ZrC_{1-x} in Table 2.22	
ZrC_{1-x}–TaC_{1-x}–$ThC_{1\pm x}$			*See* section TaC_{1-x}–$ThC_{1\pm x}$–ZrC_{1-x} in Table 2.22	
ZrC_{1-x}–TaC_{1-x}–TiC_{1-x}			*See* section TaC_{1-x}–TiC_{1-x}–ZrC_{1-x} in Table 2.22	
ZrC_{1-x}–TaC_{1-x}–$UC_{1\pm x}$			*See* section TaC_{1-x}–$UC_{1\pm x}$–ZrC_{1-x} in Table 2.22 *See also* section C–Ta–U–Zr in Table I-2.14	
ZrC_{1-x}–TaC_{1-x}–VC_{1-x}			*See* section TaC_{1-x}–VC_{1-x}–ZrC_{1-x} in Table 2.22	
ZrC_{1-x}–TaC_{1-x}–δ-$WC_{1\pm x}$			*See* section TaC_{1-x}–δ-$WC_{1\pm x}$–ZrC_{1-x} in Table 2.22	
ZrC_{1-x}–α-$Ta_{2+x}C$			*See* section α-$Ta_{2+x}C$–ZrC_{1-x} in Table 2.22 *See also* section C–Ta–Zr in Table I-2.14	
ZrC_{1-x}–β-$Ta_{2\pm x}C$			*See* section β-$Ta_{2\pm x}C$–ZrC_{1-x} in Table 2.22 *See also* section C–Ta–Zr in Table I-2.14	
ZrC_{1-x}–δ-TaN_{1-x}	N_2, 3 MPa	1800	Monocarbonitride (cubic) continuous solid solution (complete solubility in the system)	[122]
ZrC_{1-x}–ε-TaN_{1-x}	N_2, >0.1 MPa	1500	Terminal mutual solid solubilities between the components; the max. solid solubility of ε-TaN_{1-x} (hexagonal) in ZrC_{1-x} amounts to ~80 mol.%	[33, 140]
ZrC_{1-x}–$TaSi_2$	Vacuum (~0.1 kPa)	1700	The interaction between the components leads to the formation of (Zr,Ta)C_{1-x}, Zr_2Si, $ZrSi_2$ and SiC	[1423]
ZrC_{1-x}–$TbN_{1\pm x}$	–	–	Monocarbonitride (cubic) continuous solid solution (complete solubility in the system?)	[33]
ZrC_{1-x}–TcC_{1-x}	–	–	Terminal mutual solid solubilities between the components (?) *See also* section C–Tc–Zr in Table I-2.14	[947, 1409]
ZrC_{1-x}–$ThC_{1\pm x}$ (α-Th)	–	1000	Compatible with each other (practically, no solid solubility of ZrC_{1-x} in $ThC_{1\pm x}$, 120 h exposure)	[33, 200, 948–952, 1006]
	–	1100–1500	Compatible with each other (the solid solubilities of metallic Th and Th carbides in ZrC_{1-x} and those of ZrC_{1-x} in metallic Th and Th carbides are very low)	
	–	1500	ZrC_{1-x} is in equilibrium with both $ThC_{1\pm x}$ and γ-ThC_{2-x} carbides	
	–	1950–2150	The max. solid solubility of ZrC_{1-x} in $ThC_{1\pm x}$ is $\leq$3 mol.% and that of $ThC_{1\pm x}$ in ZrC_{1-x} is noticeably lower *See also* section C–Th–Zr in Table I-2.14	

(continued)

Table 5.25 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	Reference
ZrC_{1-x}–$ThC_{1\pm x}$–TiC_{1-x}	–	>2000	The mutual solubilities between ZrC_{1-x}–TiC_{1-x} monocarbide (cubic) continuous solid solution and $ThC_{1\pm x}$ are very low (the max. solubility of $ThC_{1\pm x}$ in ZrC_{1-x} is a bit higher than it is in TiC_{1-x}, those of ZrC_{1-x} and TiC_{1-x} in $ThC_{1\pm x}$ are noticeably lower)	[33, 921]
ZrC_{1-x}–$ThC_{1\pm x}$–$UC_{1\pm x}$	–	>2000	Monocarbide (cubic) continuous solid solution with great miscibility gap because of low mutual solubilities in the ZrC_{1-x}–$ThC_{1\pm x}$ system *See also* section C–Th–U–Zr in Table I-2.14	[33, 200, 948, 949]
ZrC_{1-x}–$ThC_{1\pm x}$–VC_{1-x}	–	>2000	The mutual solubilities between all the components are extremely low	[33, 921]
ZrC_{1-x}–γ-ThC_{2-x}	–	1500	Compatible with each other (mutual solid solubilities are very low) *See also* section C–Th–Zr in Table I-2.14	[33, 200, 951, 952]
ZrC_{1-x}–$ThN_{1\pm x}$	–	–	Monocarbonitride (cubic) continuous solid solution (complete solubility in the system?)	[33]
ZrC_{1-x}–$ThP_{1\pm x}$	–	–	Terminal mutual solid solubilities between the components (?)	[33]
ZrC_{1-x}–ThO_{2-x}	Vacuum 0.1 Pa	$\geq$1700	The initiation of interaction in the powdered ZrC_{1-x} and ThO_{2-x} mixtures (preliminarily cold-pressed), noticeable mass loss is observed	[42, 925, 963]
ZrC_{1-x}–ThO_{2-x}–UO_{2+x}	He flow	1850–2550	Chemical vapour deposited ZrC_{1-x} coatings (20 μm thickness) are compatible with ThO_{2-x}–UO_{2+x} solid solutions substrate (intact after heating procedure for 1 h)	[785]
ZrC_{1-x}–$ThS_{1\pm x}$	–	–	Terminal mutual solid solubilities between the components (?)	[33]
ZrC_{1-x}–TiC_{1-x}	–	1500	Monocarbide (cubic) extended solid solutions; the miscibility gap ranges in homogeneity area from ~$Zr_{0.05-0.20}Ti_{0.80-0.95}C_{0.96}$ to ~$Zr_{0.80}Ti_{0.20}C_{0.96}$ and from ~$Zr_{0.20}Ti_{0.80}C_{0.50}$ to ~$Zr_{0.60}Ti_{0.40}C_{0.56}$ (experimentally determined and theoretically calculated)	[33, 144, 193, 200, 887, 888, 890–893, 931, 953–955, 988, 1010, 1020, 1056, 1069, 1070, 1073, 1074, 1084, 1087, 1088, 1092, 1314, 1347, 1348, 1400, 1436–1438, 1443, 1444, 1447, 1448, 1476, 1484, 1485, 1533, 1536, 1537, 1539, 1540, 1541, 1546]
	–	~2000–2400	Critical point of the miscibility gap is corresponding to ~$Zr_{0.35-0.45}Ti_{0.55-0.65}C_{1-x}$ composition (experimentally determined and theoretically calculated)	
	–	~2100–2900	Monocarbide (cubic) continuous solid solution (complete solubility in the system, the variation of lattice parameter a, nm with composition for $(Zr_{1-y}Ti_y)C_{1-x}$ phases ($x \approx 0$, $0 \leq y \leq 1$) is linear: $a = 0.4700 - 0.0371y$)	
	–	2400	Sintering process of the equimolar mixture of components was studied	

(continued)

Table 5.25 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	Reference
	–	~2880–2900	Min. melting point corresponding to ~$Zr_{0.25–0.35}Ti_{0.65–0.75}C_{1–x}$ composition (experimentally determined and theoretically calculated) *See also* section C–Ti–Zr in Table I-2.14	
$ZrC_{1–x}$–$TiC_{1–x}$–δ-$TiN_{1±x}$	Vacuum, <0.1 Pa	1300–2100	Monocarbonitride (cubic) continuous solid solution (complete solubility in the system, the variation of lattice parameter a, nm with compostion for $(Zr_{1–y}Ti_y)(C_{1–x}N_x)$ phases $(0 \leq x \leq 1, 0 \leq y \leq 1)$: $a = 0.4700 - 0.0125x - 0.0371y + 0.0043xy$)	[1476, 1545]
$ZrC_{1–x}$–$TiC_{1–x}$–$UC_{1±x}$	–	2000–2050	Monocarbide (cubic) continuous solid solution with great miscibility gap because of very low mutual solubilities in the $TiC_{1–x}$–$UC_{1±x}$ system *See also* section C–Ti–U–Zr in Table I-2.14	[33, 200, 921]
$ZrC_{1–x}$–$TiC_{1–x}$–$VC_{1–x}$	–	1900–2000	Monocarbide (cubic) continuous solid solution with great miscibility gap because of very low mutual solubilities in the $ZrC_{1–x}$–$VC_{1–x}$ system	[921, 1074]
$ZrC_{1–x}$–$TiC_{1–x}$–δ-$WC_{1±x}$	–	1900–2000	Extended solid solution based on $ZrC_{1–x}$–$TiC_{1–x}$ monocarbide (cubic) continuous solid solution; the solubilities of δ-$WC_{1±x}$ in $ZrC_{1–x}$ and $TiC_{1–x}$ are ~30 and ~60 mol.%, respectively (the solubilities of $ZrC_{1–x}$ and $TiC_{1–x}$ in δ-$WC_{1±x}$ are very low)	[921, 1074]
$ZrC_{1–x}$–$TiC_{1–x}$–δ-$NbN_{1–x}$	Pure N_2	1200–2500	Monocarbonitride (cubic) continuous solid solution (complete solubility in the system)	[1532]
$ZrC_{1–x}$–δ-$TiN_{1±x}$	–	~2600	Monocarbonitride (cubic) continuous solid solution (complete solubility in the system)	[33, 46, 1476, 1484, 1545]
$ZrC_{1–x}$–$TiB_{2±x}$	–	2620–2700	Eutectic $ZrC_{1–x}$ $(x \approx 0)$ – $TiB_{2±x}$ $(x \approx 0)$	[1081, 1359, 1397, 1456]
$ZrC_{1–x}$–$TiO_{1±x}$	–	–	Terminal mutual solid solubilities between the components, ?	[33]
$ZrC_{1–x}$–$UC_{1±x}$	–	~1200–2500	Monocarbide (cubic) continuous solid solution (a spacing-composition curve in the system is almost linear, complete solubility in the system) *See also* section C–U–Zr in Table I-2.14	[33, 43, 149, 200, 260, 337, 370, 901, 927, 928, 948–950, 956–961, 1007, 1021, 1022, 1048, 1357, 1358, 1413, 1417, 1418, 1421, 1552]

(continued)

Table 5.25 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	Reference
ZrC_{1-x}–$UC_{1\pm x}$–UN_{1-x}	–	~1200–2100	Monocarbonitride (cubic) continuous solid solution *See also* section C–N–U–Zr in Table I-2.14	[1551]
ZrC_{1-x}–$UC_{1\pm x}$–VC_{1-x}	–	1900–2000	Low mutual solubilities between ZrC_{1-x}–$UC_{1\pm x}$ monocarbide (cubic) continuous solid solution and VC_{1-x}	[921]
ZrC_{1-x}–$UC_{1\pm x}$–δ-$WC_{1\pm x}$	–	1900–2000	The solubility of δ-$WC_{1\pm x}$ in ZrC_{1-x}–$UC_{1\pm x}$ monocarbide (cubic) continuous solid solution increases with increasing content of Zr from very low values up to ~30 mol.%	[33]
ZrC_{1-x}–$UC_{1\pm x}$–α/β-UC_{2-x}	–	1700–2000	The composition of ZrC_{1-x}–$UC_{1\pm x}$ monocarbide (cubic) continuous solid solution in the equilibrium with UC_{2-x}, shifts from ~$Zr_{0.60-0.65}U_{0.35-0.40}C_{1-x}$ (at 1700 °C) to ~$Zr_{0.70-0.80}U_{0.20-0.30}C_{1-x}$ (at 1900–2000 °C)	[901, 958, 961]
	–	~2400	Quasi-eutectic $(Zr,U)C_{1-x}$–β-$(U,Zr)C_{2-x}$–C *See also* section C–U–Zr in Table I-2.14	
ZrC_{1-x}–$UC_{1\pm x}$–$US_{1\pm x}$	–	1650	The solubility of $US_{1\pm x}$ in ZrC_{1-x}–$UC_{1\pm x}$ monocarbide (cubic) continuous solid solution is ~5 mol.% (independent on $(Zr,U)C_{1\pm x}$ composition); solubilities of ZrC_{1-x} and $UC_{1\pm x}$ in $US_{1\pm x}$ are ~15 and ~40 mol.%, respectively	[1352]
	–	2100	The solubility of $US_{1\pm x}$ in ZrC_{1-x}–$UC_{1\pm x}$ monocarbide (cubic) continuous solid solution is ~7.5 mol.% (independent on $(Zr,U)C_{1\pm x}$ composition); solubilities of ZrC_{1-x} and $UC_{1\pm x}$ in $US_{1\pm x}$ are ~18 and ~40 mol.%, respectively	
	–	>2100	The solubility of $US_{1\pm x}$ in ZrC_{1-x}–$UC_{1\pm x}$ monocarbide (cubic) continuous solid solution is ~10–20 mol.%; solubilities of ZrC_{1-x} and $UC_{1\pm x}$ in $US_{1\pm x}$ are ~21 and ~42 mol.%, respectively	
ZrC_{1-x}–β-UC_{2-x}	–	~2400	Quasi-eutectic ZrC_{1-x}–β-UC_{2-x}; the max. solubility of Zr in β-UC_{2-x} is corresponding to ~$(U_{0.94}C_{0.06})C_{1.94}$ composition *See also* section C–U–Zr in Table I-2.14	[33, 901, 961]
ZrC_{1-x}–UN_{1-x}	–	–	Monocarbonitride (cubic) continuous solid solution (complete solubility in the system)	[33, 798]
ZrC_{1-x}–$UP_{1\pm x}$	–	–	Terminal mutual solid solubilities between the components (?)	[33]
ZrC_{1-x}–UO_{2+x}	–	1500	Chemical vapour deposited ZrC_{1-x} coatings are compatible with UO_{2+x}	[1211, 1225]

(continued)

Table 5.25 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	Reference
ZrC_{1-x}–UO_{2+x}–ThO_{2-x}	He flow	1850–2550	Chemical vapour deposited ZrC_{1-x} coatings (20 μm thickness) are compatible with UO_{2+x}–ThO_{2-x} solid solution substrate (intact after heating procedure for 1 h)	[785]
ZrC_{1-x}–$US_{1\pm x}$	–	1300	The max. solid solubility of ZrC_{1-x} in $US_{1\pm x}$ is ~13 mol.% and that of $US_{1\pm x}$ in ZrC_{1-x} is ~3 mol.%	[33, 1004, 1352, 1353]
	–	2470	The following peritectic reaction is observed: L + ZrC_{1-x} ($US_{1\pm x}$) *solid solution* → $US_{1\pm x}$ (ZrC_{1-x}) *solid solution* with the max. solid solubility of ZrC_{1-x} in $US_{1\pm x}$ – 21.5–22.5 mol.% and that of $US_{1\pm x}$ in ZrC_{1-x} – ~19 mol.%	
ZrC_{1-x}–VC_{1-x}	Ar, >0.1 MPa	1900	Additions of VC_{1-x} (3.5–13.5 mol.%), higher than the max. solid solubility of VC_{1-x} in ZrC_{1-x} (1.8 mol.%), effectively promote the hot-pressing densification process (up to nearly poreless state)	[33, 38, 144, 193, 200, 888, 953, 1061, 1062, 1070, 1074, 1078, 1539]
	–	2100	Terminal monocarbide (cubic) solid solutions; the solid solubility of ZrC_{1-x} in VC_{1-x} is ~1.0 mol.% and that of VC_{1-x} in ZrC_{1-x} is ~5.0 mol.%	
	–	~2460–2530	Eutectic VC_{1-x} (x = 0.12) – ZrC_{1-x} ($x \approx 0$); the solid solubility of ZrC_{1-x} in VC_{1-x} is ~2.5–5.0 mol.% and that of VC_{1-x} in ZrC_{1-x} is ~8.0–8.5 mol.% *See also* section C–V–Zr in Table I-2.14	
ZrC_{1-x}–VC_{1-x}–δ-$WC_{1\pm x}$	–	~1900–2000	The max. solubility of δ-$WC_{1\pm x}$ in cubic carbides ZrC_{1-x} and VC_{1-x} corresponds to ~$(Zr_{0.3}W_{0.7})C_{1\pm x}$ and ~$(V_{0.4}W_{0.6})C_{1-x}$ compositions, respectively; the max. mutual solubilities of carbides ZrC_{1-x} and VC_{1-x} corresponds to ~$(Zr_{0.95}V_{0.05})C_{1-x}$ and ~$(V_{0.99}Zr_{0.01})C_{1-x}$ compositions, respectively, and with increasing V content in $(Zr,V)C_{1-x}$ the max. solubility of W decreases there	[33, 921, 1074]
ZrC_{1-x}–δ-VN_{1-x}	–	–	Terminal solid solutions with negligible mutual solubilities (?)	[33, 46]
ZrC_{1-x} – γ-WC_{1-x}	–	~2500–2800	Monocarbide (cubic) continuous solid solution *See also* section C–W–Zr in Table I-2.14	[33, 200, 896, 899, 926, 1093]

(continued)

Table 5.25 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	Reference
ZrC_{1-x}– δ-$WC_{1\pm x}$	–	1500	No solubility of Zr in δ-$WC_{1\pm x}$	[33, 200, 373, 871, 882, 894, 899, 901, 1046, 1093, 1380, 1479]
	–	1600	The max. solid solubility of W in ZrC_{1-x} corresponds to ~$(Zr_{0.80–0.85}W_{0.15–0.20})C_{1-x}$ composition	
	–	1950–2000	The max. solid solubility of W in ZrC_{1-x} corresponds to ~$(Zr_{0.60–0.70}W_{0.30–0.40})C_{1-x}$ composition	
	–	≥2100	The max. solid solubility of W in ZrC_{1-x} corresponds to $(Zr_{0.04–0.40}W_{0.60–0.96})C_{1-x}$ composition *See also* section C–W–Zr in Table I-2.14	
ZrC_{1-x}– $\alpha/\beta/\gamma$-$W_{2\pm x}C$	–	–	The max. solid solubility of Zr in γ-$W_{2\pm x}C$ is ~3.5 at.% *See also* section C–W–Zr in Table I-2.14	[33, 899, 1093]
ZrC_{1-x}– β-W_2B_{5-x}	–	2240	Eutectic ZrC_{1-x}–β-W_2B_{5-x}	[1399]
ZrC_{1-x}–$YC_{1\pm x}$	–	–	Monocarbide (cubic) continuous solid solution (complete solubility in the system?)	[33]
ZrC_{1-x}–$YN_{1\pm x}$	–	–	Monocarbonitride (cubic) continuous solid solution (complete solubility in the system?)	[33]
ZrC_{1-x}–$YbN_{1\pm x}$	–	–	Monocarbonitride (cubic) continuous solid solution (complete solubility in the system?)	[33]
ZrC_{1-x}–$ZrN_{1\pm x}$	–	1100–2000	Monocarbonitride (cubic) continuous solid solution (complete solubility in the system, the variation of lattice parameter a, nm with compostion for $Zr(C_{1-x}N_x)$ phases ($0 \leq x \leq 1$) is linear: $a = 0.4700 - 0.0125x$) with the homogeneity range within ~$ZrC_{0.56–0.59}$ – ~$ZrC_{0.95}$ – ~$ZrN_{0.54–0.56}$ – ~$ZrN_{1.04}$ compositions *See also* section C–N–Zr in Table I-2.14	[33, 46, 51, 59, 110, 124, 126, 200, 278, 328, 472, 576, 703, 962, 992, 994, 1201, 1434, 1446, 1476, 1483, 1484]
ZrC_{1-x}–$ZrN_{1\pm x}$– $ZrB_{2\pm x}$	–	–	Quasi-binary eutectic system $Zr(C,N)_{1-x}$–$ZrB_{2\pm x}$	[1079]
ZrC_{1-x}–$ZrN_{1\pm x}$– 'ZrO'	–	–	Formation of $ZrC_{1-x}N_yO_z$ oxycarbonitride phase[a] based on of monocarbonitride (cubic) ZrC_{1-x}–$ZrN_{1\pm x}$ continuous solid solution	[44, 124, 165, 962, 994, 1027, 1415]
	–	1600	Homogeneity range limits of $ZrC_xN_yO_z$ phase are ~$ZrC_{0.45}O_{0.55}$ and ~$ZrN_{0.65}O_{0.35}$	
	–	2000	Homogeneity range limits of $ZrC_xN_yO_z$ phase are ~$ZrC_{0.25}O_{0.75}$ and ~$ZrN_{0.45}O_{0.55}$ *See also* section C–N–O–Zr in Table I-2.14	
ZrC_{1-x}–$ZrP_{1\pm x}$	–	–	Terminal mutual solid solubilities between the components, ?	[33]

(continued)

Table 5.25 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	Reference
ZrC_{1-x}–$ZrB_{2\pm x}$	–	2360	Eutectic ZrC_{1-x}–$ZrB_{2\pm x}$–C	[33, 200, 840, 842, 843, 845, 967–969, 995, 1008, 1012, 1081, 1359, 1397, 1416, 1453–1459, 1462, 1544]
	–	~2620–2850	Eutectic ZrC_{1-x} (x = 0.03–0.12)–$ZrB_{2\pm x}$ ($x \approx 0$); the max. solid solubility of $ZrB_{2\pm x}$in ZrC_{1-x} is ~1–5 mol.% and that of ZrC_{1-x} in $ZrB_{2\pm x}$ is <2 mol.%	
	–	2815–2845	Eutectic ZrC_{1-x} ($x \approx 0$)–$ZrB_{2\pm x}$ ($x \approx 0$)	
			See also section C–B–Zr in Table I-2.14	
ZrC_{1-x}–β/γ-ZrO_{2-x}	Vacuum, 0.1–1.0 Pa	1400–1450	The initiation of interaction in the powdered ZrC_{1-x} and β/γ-ZrO_{2-x} mixtures (preliminarily cold-pressed)	[14, 33, 42, 44, 63, 74, 124, 130, 151, 161–164, 230, 923, 924, 963–966, 970–979, 986, 997–1001, 1017, 1018, 1027, 1028, 1075, 1077, 1154, 1170, 1171, 1197–1204, 1392, 1393, 1396, 1402, 1415, 1432, 1433, 1445, 1452, 1513]
	–	1500–2000	Formation of $ZrC_{1-x}O_y$ oxycarbide (cubic) phases, extended solid solutions based on ZrC_{1-x} with ~$ZrC_{0.60}$ – ~$ZrC_{0.50}O_{0.20}$ – ~$ZrC_{0.40-0.60}O_{0.30-0.55}$ – ~$ZrC_{0.80}O_{0.20}$ – $ZrC_{0.99}$ approximate homogeneity limits	
	Vacuum, 1.0 Pa	≥1800	Formation of intermediate ZrC_yO_δ oxycarbide (cubic) phases (with shortened lattice parameter) in the powdered mixtures of components in accordance with the following reactions: $aZrC_{1-x} + bZrO_{2-z} = (a + b)ZrC_yO_\delta +$ $+ [a(1 - x) - y(a + b)]CO\uparrow$, where $\delta = [b(2 - z) + y(a + b) - a(1 - x)]/(a + b)$, or in experimental practice: $4ZrC + ZrO_2 = 5ZrC_{0.7}O_{\sim 0.3} + ½CO$ (determined for powdered mixture 80 mol.% $ZrC_{\sim 1.0}$–20 mol.% $ZrO_{\sim 2.0}$ vacuum heat treatment, resulting in the formation of single-phase products)	
	Vacuum, 0.1–1.0 Pa	2000	In the equimolar powdered ZrC_{1-x} and β/γ-ZrO_{2-x} mixture carbon content loss (difference between initial and final carbon contents) amounts to 2% (2 h exposure)	
	Vacuum, 0.1 Pa	2050–2150	In the conditions of high vacuum and long-term exposures (up to 24 h), the formation of metallic solid solutions α-Zr(C,O) with $ZrC_{0.02-0.03}$ and $ZrO_{0.41-0.54}$ solubility limits was observed	
	–	>2200	The initiation of interaction in the contact zone between compact dense materials (formation of intermediate phases)	
	–	~2700	ZrC_{1-x} materials are chemically resistant to molten ZrO_{2-x}	

(continued)

Table 5.25 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	Reference
	Ar (p_{O2} ≈ 1 Pa)	≥3300	$ZrC_{0.82}C_{0.14}$ surface heating leads to $ZrC_{0.75}$ formation; O releases in the form of the gaseous species ZrO (?), while the grains of oxycarbide, converted into the carbide, grew from 2 to 20 μm	
	–	–	Formation of carbon-stabilized (cubic) γ-$ZrO_{2-x}C_z$ oxide (solid solution, ?) and ZrO_2C_x phase (?) *See also* section C–O–Zr in Table I-2.14	

[a]Practically, due to the serious difficulties in manufacturing carbides non-contaminated in any degree by O and N, which are dissolved at higher temperatures easily, all the materials labeled as ZrC_{1-x} would have to be considered as $ZrC_{1-x}N_yO_z$ with low or very low values of y and z indexes

Table 5.26 Chemical interaction of zirconium carbide phases with gaseous media at elevated, high and ultra-high temperatures (reaction systems are given in alphabetical order)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	Reference
ZrC_{1-x}–Br_2	–	20	No interaction with Br_2 dissolved in CCl_4 (200 h exposure)	[452, 1331]
	Br_2	>500	Reacts readily	
ZrC_{1-x}–CO[a]	CO	–	Formation of $ZrC_{1-x}O_y$ oxycarbide phases (extended substitution solid solution based on ZrC_{1-x} with approximate homogeneity limits: ~$ZrC_{0.60}$–~$ZrC_{0.50}O_{0.20}$–~$ZrC_{0.40–0.60}O_{0.30–0.55}$–~$ZrC_{0.80}O_{0.20}$–~$ZrC_{0.99}$)	[10, 33, 44, 74, 124, 1001, 1131, 1199, 1200, 1204, 1209–1211, 1342, 1343, 1402, 1445, 1513]
	CO	1150–1250	For the "imaginary" reaction $ZrC_{\sim1.0}$ + 2CO (gas) ↔ $ZrO_{\sim2.0}$ + 3C (graphite) (not taking into account the formation of oxycarbide phases) near-equilibrium pressure p_{CO}, Pa measured experimentally by torsion effusion method is obeyed the following rule lg p_{CO} = 11.55 − 16,100/T, where T is temperature, K	
	CO	1600–1950	For the "imaginary" reaction $ZrC_{\sim1.0}$ + 2CO (gas) ↔ $ZrO_{\sim2.0}$ + 3C (graphite) (not taking into account the formation of oxycarbide phases) near-equilibrium pressure p_{CO}, Pa measured experimentally by manometric method is obeyed the following rule lg p_{CO} = 13.59 − 16,580/T, where T is temperature, K *See also* section C–O–Zr in Table I-2.14	

(continued)

Table 5.26 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	Reference
ZrC_{1-x}–CO_2[a]	CO_2	–	Formation of $ZrC_{1-x}O_y$ oxycarbide phases (extended substitution solid solution based on ZrC_{1-x} with approximate homogeneity limits: ~$ZrC_{0.60}$–~$ZrC_{0.50}O_{0.20}$–~$ZrC_{0.40–0.60}O_{0.30–0.55}$–~$ZrC_{0.80}O_{0.20}$–~$ZrC_{0.99}$) and subsequent formation of oxide scales	[33, 44, 74, 124, 1001, 1131, 1152, 1199, 1200, 1209, 1210, 1402, 1513]
	CO_2	–	The selective character of interaction is inherent for ZrC_{1-x} materials, it is described by the following reactions: $ZrC_{1-x} + yCO_2 = ZrC_{1-x-y} + 2yCO\uparrow$ (for $x \approx 0$, accompanied with mass loss), $ZrC_{1-x} + yCO_2 = ZrO_{2-z} + (3 - x - y - z)C + + (2y + z - 2)CO\uparrow$ (for $x >> 0$, accompanied with mass gain) *See also* section C–O–Zr in Table I-2.14	
ZrC_{1-x}–Cl_2	Cl_2	~0–250	The preferential reaction under equilibrium conditions (in accordance with thermodynamic analysis): $ZrC_{1-x} + (4 - 2x)Cl_2 = ZrCl_4 + (1 - x)CCl_4\uparrow$	[32, 35, 46, 223, 452, 1130, 1132, 1133, 1135–1137]
	Cl_2	~200–250	The initiation of interaction between powdered ZrC_{1-x} and gaseous Cl_2 observed experimentally	
	Cl_2	>250	Powdered ZrC_{1-x} burns in Cl_2 or Cl_2-rich atmosphere	
	Cl_2	~250–750	The interaction leads to the volatilization of $ZrCl_4$ and formation of carbon: $ZrC_{1-x} + (3 - x)Cl_2 = ZrCl_4\uparrow + [(1 - x)/2]C + [(1 - x)/2]CCl_4\uparrow$ via the chemical reaction, which is preferential under equilibrium conditions (in accordance with thermodynamic analysis)	
	Cl_2	400	Powdered ZrC_{1-x} is easily attacked by Cl_2	
	Cl_2	450–650	The chlorination rate is linear (apparent activation energy $E \approx 60$ kJ mol^{-1})	
	Cl_2	550	ZrC_{1-x} powder (850 mg) are chlorinated completely after 4 h exposure[b]	
	Cl_2	650–950	The chlorination rate is linear (apparent activation energy $E \approx 10$ kJ mol^{-1})	
	Cl_2	700–800	Powdered ZrC_{1-x} decomposes easily and completely	
	Cl_2	~700–1200	The preferential reaction under equilibrium conditions (in accordance with thermodynamic analysis): $ZrC_{1-x} + 2Cl_2 = ZrCl_4\uparrow + (1 - x)C$; in steady-flow chlorination, the gas flow rate was shown to have a significant influence on the reaction rate, a decrease in Cl_2 gas pressure creates a local atmosphere near the ZrC_{1-x} surface causing the reaction rate to drop	

(continued)

Table 5.26 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	Reference
ZrC_{1-x}–F_2	F_2	>25	Powdered ZrC_{1-x} burns in F_2 or F_2-rich atmosphere	[46, 223, 452, 1134, 1136, 1137, 1207]
	F_2	~250–300	Electron-beam melted $ZrC_{\sim 1.0}$ burns in F_2 or F_2-rich atmosphere	
	F_2 (4.1 kPa), He	280–410	The formation of soft and fluffy uniform ZrF_4 scales with poor adhesion, the reaction rate is linear (mass gain with 0.1–3.7 mg cm^{-2} min^{-1}, $E = 92 \pm 7$ kJ mol^{-1}, the rate is approximately proportional to $p_{F2}^{1/2}$)	
	F_2 (0.4 kPa), He	720–970	Since ZrF_4 is very volatile in this temperature range, practically no scales is formed on ZrC_{1-x} surface, the reaction rate is linear (mass loss with 0.91–1.05 mg cm^{-2} min^{-1}, practically $E \approx 0$, the rate is approximately proportional to $p_{F2}^{3/2}$), the rate-determining process is the diffusion of F to the reaction surface through the gaseous products	
ZrC_{1-x}–H_2	H_2	–	Coefficients of surface recombination of H atoms (measured experimentally by "mobile overlapping sample" technique) are 0.056 for $ZrC_{0.80}$ and 0.030 for $ZrC_{0.99}$ phases	[33, 223, 260, 319, 336, 337, 452, 515, 517, 762, 1021, 1022, 1138–1152, 1249]
	H_2	–	Formation of carbohydride phases ZrC_xH_y (x = 0.30–0.65, y = 0.30–1.25, including Zr_2CH_x and Zr_3CH_x)[c]	
	H_2 (pure)	800	The total content of carbon in powdered $ZrC_{0.97}$ after 2 h exposure decreases from 11.5% to 11.1% (content of non-combined carbon from 0.1% drops to zero)	
	H_2 (flow)	1250	Powdered ZrC_{1-x} interacts (or burns) intensively	
	H_2, C_2H_2	1730–2730	For the following reaction $ZrC_{1-x}(solid) + [(1 - x)/2]H_2(gas) = Zr(gas) + [(1 - x)/2]C_2H_2(gas)$, calculated equilibrium partial pressures p_{C2H2} are 10^{-9}–5×10^{-3} Pa for $ZrC_{0.60}$ and 50–2×10^4 Pa for $ZrC_{0.99}$	
	H_2, CH_4	1730–2730	For the following reaction $ZrC_{1-x}(solid) + 2(1 - x)H_2(gas) = Zr(gas) + (1 - x)CH_4(gas)$, calculated equilibrium partial pressures p_{CH4} are 10^{-4}–10^{-2} Pa for $ZrC_{0.60}$ and 50–10 Pa for $ZrC_{0.99}$	
	H_2	2000–2500	Weak interaction with the minimal determined mass change	
	H_2	2200–2480	Powdered ZrC_{1-x} is stable in static conditions (2 h exposure)	

(continued)

Table 5.26 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	Reference
	H_2	~2800	The exposure of ZrC_{1-x} materials in H_2 for 1–2 h results in inhomogeneous carbon concentration in bulk, but also leads to the change of microstructure, decrease of density and five-fold drop of strength	
	H_2	~3100	The mass loss of $ZrC_{\sim 1.0}$ materials amounts up to ~4 × 10^{-2} g cm^{-2} (1⅔ h exposure) *See also* section C–H–Zr in Table I-2.14	
ZrC_{1-x}–'ZrO'–H_2	H_2	–	Formation of $ZrO_{0.6}C_{0.2}H_x$ *See also* section C–H–O–Zr in Table I-2.14	[1138]
ZrC_{1-x}–H_2O[a]	H_2O, 100 MPa	200–400	The hydrothermal oxidation reaction $ZrC_{1-x} + (2-y)H_2O = ZrO_{2-y} + (1-x)CH_4 + (2x-y)H_2$ of powdered $ZrC_{0.98}$ (mean particle size – 3.2 μm, contents: non-combined C – 0.24%, O – 0.11%, N – 0.25%, H – 0.002%, Fe – 0.01%) converts it into α-ZrO_{2-x} (completely at >270 °C and exposure >10 h, or at >340 °C and exposure >3 h); the process is controlled by phase boundary reactions with apparent activation energy $E \approx$ 80–100 kJ mol^{-1} (protective oxide layer does not form)	[10, 33, 46, 1131, 1138, 1155–1157, 1193, 1210]
	H_2O	300–400	Bulk ZrC_{1-x} materials are stable in water steam	
	H_2O	315	Non-stoichiometric ZrC_{1-x} phases are more reactive than near-stoichiometric phase	
	H_2O/O_2. 21–42/ >0.5 kPa	450–580	Water vapour accelerates the oxidation surface reaction of powdered ZrC_{1-x}	
	H_2O, He	500–950	The main hydrolytic reactions in superheated steam are: $ZrC_{1-x} + 2(1-x)H_2O = ZrO_{2-2x} + + (1-x)CH_4$, $ZrC_{1-x} + (4-b)H_2O = ZrO_{2-x(b-2)} + + b(1-x)CO + (1-b)(1-x)CO_2 + + (4-b)H_2$, while the contribution of the reaction $ZrC_{1-x} + 2(1-x)H_2O = ZrO_{2-2x} + (1-x)C + + 2(1-x)H_2$ is rather negligible; powdered $ZrC_{0.99}$ (contents: non-combined C – 0.28%, O – 0.25%, N – 0.19%) converts into β-ZrO_{2-x} at ≤800 °C and α-ZrO_{2-x} – at ≥800 °C, with temperature growth the evolution of H_2 increases monotonously from 7.1 to 28.5 mol kg^{-1} and molar ratio of $CO:CO_2:CH_4$ (a conversion rate is given in brackets) varies from 1.0:0.5:0.6 (2.1%) at 500 °C to 1.0:2.5:0.03 (77.2%) at 800 °C, and 1.0:0.7:0.004 (93.3%) at 950 °C	

(continued)

Table 5.26 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	Reference
	H_2O	–	Formation of $ZrO_{0.6}C_{0.2}H_x$ oxycarbohydride phases *See also* section C–H–O–Zr in Table I-2.14	
ZrC_{1-x}–$ZrN_{1\pm x}$ –H_2O	H_2O, He	500–950	Hydrolysis of ZrC_xN_y monocarbonitride phases leads to the evolution of H_2 (larger amounts of H_2 evolves in the phases richer in C), CO, CO_2 (delayed in release), CH_4 (small amounts, the phases richer in N are more productive of it), NH_3 and N_2, and precipitation of C (converted thereafter into CO_2 and H_2); the residue consists of α/β-ZrO_{2-x} mixture at lower temperatures and α-ZrO_{2-x} phase at higher temperatures	[1156]
ZrC_{1-x}–HX (X = F, Cl, Br, I)	HX	–	Powdered ZrC_{1-x} decomposes by gaseous hydrogen halides easily	[223]
ZrC_{1-x}–He			*See* Tables 5.27 and 5.29	
ZrC_{1-x}–I_2	I_2	500–800	Powdered ZrC_{1-x} burns in I_2 or I_2-rich atmosphere	[223, 452, 1354]
	I_2	780–800	The ZrI_4 yield of the following reaction $ZrC_{1-x} + 2I_2 = ZrI_4 + (1-x)C$ is 97% (15 h exposure)	
ZrC_{1-x}–Kr			*See* Table 5.27	
ZrC_{1-x}–N_2	N_2	>800	Formation of various monocarbonitride ZrC_xN_z ($0.54 \leq x+z \leq 1.04$) and mononitride $ZrN_{1\pm x}$ phases	[1, 32, 33, 39, 42, 46, 47, 51, 59, 124, 126, 200, 213, 223, 278, 691, 692, 703, 962, 992, 994, 1153, 1158, 1201]
	N_2	1150	Formation of monocarbonitride (cubic) continuous solid solution with the homogeneity range within ~$ZrC_{0.56-0.59}$ – ~$ZrC_{0.95}$–~$ZrN_{0.54-0.56}$–~$ZrN_{1.04}$ compositions	
	N_2, 0.1–30 MPa	1400	Treatment of porous ZrC_{1-x} in N_2 leads to the formation of $ZrC_{0.48}N_{0.52}$ at p_{N2} = 0.1 MPa and $ZrC_{0.28}N_{0.72}$ at p_{N2} = 30.0 MPa (36–37 h exposure)	
	N_2	>1500	Formation of mononitride $ZrN_{1\pm x}$ phases	
	N_2	1500–2500	Formation of various monocarbonitride ZrC_xN_z and mononitride $ZrN_{1\pm x}$ phases	
	N_2, 0.1–30 MPa	1800	Treatment of porous ZrC_{1-x} in N_2 leads to the formation of $ZrC_{0.69}N_{0.31}$ at p_{N2} = 0.1 MPa and $ZrC_{0.39}N_{0.61}$ at p_{N2} = 30.0 MPa (8 h exposure) Data available in literature are controversial *See also* section C–N–Zr in Table I-2.14	

(continued)

Table 5.26 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	Reference
ZrC_{1-x}–NH_3	NH_3	<400–500	No interaction	[223]
ZrC_{1-x}–O_2[a,d,e]	Air, O_2	–	Formation of $ZrC_{1-x}O_y$ oxycarbide (cubic) phases, extended solid solutions based on ZrC_{1-x} with ~$ZrC_{0.60}$–~$ZrC_{0.50}O_{0.20}$–~$ZrC_{0.40–0.60}O_{0.30–0.55}$–~$ZrC_{0.80}O_{0.20}$–$ZrC_{0.99}$ approximate homogeneity limits	[1, 14, 30–33, 35, 37, 42, 46, 74, 92, 96, 106–108, 124, 223, 250, 265, 319, 452, 519, 630, 932, 971, 972, 977, 980–983, 986, 999–1001, 1075, 1077, 1152, 1154, 1159–1224, 1248, 1264, 1265, 1322–1327, 1332, 1373, 1383, 1402, 1423, 1424, 1445, 1451, 1452, 1551]
	O_2, 10^{-8} Pa	20–30	The single crystal $ZrC_{0.93}$ (111) clean surface (prepared in a vacuum chamber by flashing at T > 1500 °C) adsorbs O_2 dissociatively (chemisorption bonding)	
	O-plasma, 6.7 Pa	20–30	No oxidation of ZrC_{1-x} coating and powders (exposure – up to 4 h)	
	Air	250–400	During the heat treatment of $ZrC_{0.60}$ the diffusion of O is significantly facilitated through the ordered C vacancies with the formation of ordered $ZrC_{0.60}C_{0.40}$ oxycarbide; at > 350 °C metastable β-ZrO_{2-x} nanocrystals are gradually developed	
	Air	200–450	The oxidation of hot-pressed $ZrC_{\sim 1.0}$ leads to the formation of oxycarbide $ZrC_{1-x}O_x$ outer layer, which keeps little change with time while x attains 0.58; the electrical resistance of outer layer increases from ~0.4 to ~7 μΩ m	
	Air	280–290	TG-curves and DTA-peaks indicate onset of oxidation of nanopowdered $ZrC_{\sim 1.0}$ (mean grain size – 40 nm)	
	Flow O_2 (5–50 kPa), Ar	320–1000	The non-isothermal oxidation kinetics of powdered $ZrC_{\sim 1.0}$ (surface area 0.7 $m^2\ g^{-1}$) (with heating rates varied from 5 to 15 K min^{-1}) is described by Ginstling-Brounstein's equation with apparent activation energies $E \approx 140$ kJ mol^{-1} (p_{O2} = 5 kPa), $E \approx$ 110–125 kJ mol^{-1} (p_{O2} = 10 kPa) and $E \approx 100$ kJ mol^{-1} (p_{O2} = 50 kPa); the conversion of carbide to oxides is controlled by O diffusion through the formation of oxycarbide as the intermediate, γ-ZrO_{2-x} (stabilized by C) and α-ZrO_{2-x} are present in the oxide scales at T < 800 °C, while only α-ZrO_{2-x} is in the scales at T > 800 °C	
	O_2 (1–40 kPa), Ar	380–400	TG-curves and DTA-peaks indicate onset of oxidation of powdered $ZrC_{\sim 1.0}$ (grain size – 1–10 μm, surface area – 1.7 $m^2\ g^{-1}$, contents: non-combined C < 0.5%, O – ~2.0%)	

(continued)

Table 5.26 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	Reference
	O_2/Ar, 1.3–7.9/ 32–38 kPa	380–550	The oxidation kinetics of powdered $ZrC_{\sim 1.0}$ (mean grain size – 3.3 μm, surface area – 0.7 $m^2\ g^{-1}$) is described by Jander's equation with apparent activation energy $E \approx$ 140 kJ mol^{-1} (<470 °C) and $E \approx$ 180 kJ mol^{-1} (>470 °C); the oxidation process is determined by O diffusion through compact quasi-amorphous γ-ZrO_{2-x} layer (<470 °C), or faster O diffusion along the grain boundaries in crystallized γ-ZrO_{2-x} (>470 °C)	
	Air (circulated)	385	A TG-curve indicates onset of oxidation (mass gain) of powdered $ZrC_{0.70}$ (mean particle size – 70–80 μm)	
	O_2, 1.3 Pa –2.7 kPa	400–600	Oxidation of sintered $ZrC_{0.86}$ (porosity – 5 and 30%, contents: non-combined C – 0.1%, O $\leq$ 0.05%, N $\leq$ 0.05%) leads to the increase of content of non-combined C, it is dispersed in $ZrC_{1-x}O_y$ solid solutions (at 400–500 °C) and in γ-ZrO_{2-x} (at 600 °C)	
	Air (circulated)	400–600	The formation of ZrO_2C_x phase (?) with interplanar spacings slightly different from those of β-ZrO_{2-x}	
	Air (flow)	400–900	Powdered ZrC_{1-x} (mean particle size – 1.5 μm, specific surface area – 0.59 m^{-2} g) is oxidized 22% at 400 °C, 90% at 500 °C and complete at >500 °C, no change in the surface area at 400 °C, but the max. increase is up to 15.2 m^{-2} g at 500 °C (or from 16.7 m^{-2} g after 0.25 h to 3.2 m^{-2} g after 20 h) and smaller increases are 9.9 m^{-2} g (at 600 °C) and 5.3 m^{-2} g (at 700 °C) due to sintering of the oxide scales, the data not indicated specially are for 1 h exposure (E = 14.6 ± 0.3 kJ mol^{-1}); the oxidation products are oxycarbide/suboxide intermediates (only at 400–500 °C and in traces, with a new phase (111) topotactically produced from (200) of initial carbide lattice), γ-ZrO_{2-x} (at 400–600 °C), γ-ZrO_{2-x} and α-ZrO_{2-x} (at >600 °C, with increasing amounts of α-ZrO_{2-x} at higher temperatures)	
	O_2, 13 kPa	400–1000	The effect of temperature on the oxidation of ZrC_{1-x} is estimated by activation energy $E \approx$ 290 kJ mol^{-1} for diffusion controlled mechanism (at lower temperatures) and $E \approx$ 190 kJ mol^{-1} for phase boundary controlled mechanism (at higher temperatures)	
	Air	450	The oxidation mass gain of $ZrC_{\sim 1.0}$ is ~20 mg cm^{-2} (1 h exposure) and ~60 mg cm^{-2} (2 h exposure)	

(continued)

Table 5.26 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	Reference
	O_2, 6.5 Pa – 0.7 MPa	450–580	Oxidation kinetics of powdered ZrC_{1-x} is controlled by two parallel processes: surface reaction of γ-ZrO_{2-x} formation (mostly accounted, kinetics is directly proportional to p_{O2} (at p_{O2} <10 kPa) and described by linear law, apparent activation energy $E \approx 190$ kJ mol^{-1}) and O diffusion forming oxycarbide solid solution $ZrC_{1-x}O_y$ ($E \approx 220$ kJ mol^{-1}); the initial process of O chemisorption on ZrC_{1-x} surface is characterized by $E = 54$ kJ mol^{-1}	
	Flow O_2, 0.13 Pa–13 kPa	450–700	The oxidation of powdered ZrC_{1-x} leads to the formation of γ-ZrO_{2-x} stabilized by the presence of C; no oxycarbide phases were detected	
	Air	450–800	The oxidation of ZrC_{1-x} powders is detectable from 450 °C, both α-ZrO_{2-x} and β-ZrO_{2-x} phases cooexist in the oxide scales at 550–800 °C; β-ZrO_{2-x} is stabilized at lower temperatures on smaller particles, at $T > 650$ °C the formation of α-ZrO_{2-x} is favoured	
	O_2	~480	A TG-DSC-curve indicates onset of oxidation (mass gain) of fine-dispersed $ZrC_{\sim 1.0}$ (mean particle size – 0.8 μm, specific surface area – 1.3 m^2 g^{-1})	
	Flow O_2 (20 kPa), Ar	485–530	The isothermal oxidation kinetics of powdered $ZrC_{\sim 1.0}$ (surface area – 0.7 m^2 g^{-1}) is described by Ginstling-Brounstein equation with activation energy $E \approx 130$ kJ mol^{-1}	
	O_2 (1–40 kPa), Ar	500–540	MS-curves indicate onset of C burning-off (CO_2 evolution) in powdered $ZrC_{\sim 1.0}$ (grain size – 1–10 μm, surface area – 1.7 m^2 g^{-1}, contents: non-combined C < 0.5%, O – ~2.0%); the current rate of oxidation (dependent on p_{O2}), defined for the complete carbide-oxide conversion, reaches 20–40%	
	O_2, 6.7 Pa –1.3 kPa	500–600	At earlier stages (0.5–30 min exposure) the oxidation of ZrC_{1-x} phases is accompanied with release of CO (for $x \leq 0.03$) or a CO_2 + CO mixture with ratio CO_2 : CO > 1 (for $x \geq 0.23$, the ratio increases with growth of value x) in accordance with the following reactions: $ZrC_{1-x} + y/2\ O_2 = ZrC_{1-x-y} + yCO\uparrow$ (for $x \leq 0.03$, accompanied with mass loss), $ZrC_{1-x} + (2-z)/2\ O_2 = ZrO_{2-z} + (1-x)C$ $C + (m + n/2)O_2 = mCO_2\uparrow + nCO\uparrow$ (for $x \geq 0.23$, $m/n > 1$, accompanied with mass gain)	

(continued)

Table 5.26 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	Reference
	O_2 (2.6 kPa), Ar	500–600	The oxidation of single crystal $ZrC_{0.97}$ (100) leads to the formation of the oxide scale divided into inner (poreless amorphous ZrO_{2-x} containing γ-ZrO_{2-x} particles in the regions remote from the reaction surface and 14–23 at.% of residual (unoxidized) C) and outer (aggregated γ-ZrO_{2-x} particles containing 7–10 at.% of residual C) zones; the thickness of inner zone increases parabolically reaching a constant value of 2–3 μm (at 500 °C, 240 h exposure) in contrast to that of outer zone, which increases linearly with time	
	Air	500–850	In the initial stage of the oxidation of $ZrC_{0.95}$ powder the effect of temperature is estimated by $E \approx 170$ kJ mol^{-1}	
	Air	500–850	The oxidation of powdered $ZrC_{0.98}$ (mean particle size – 10–40 μm, contents: non-combined C – 0.30%, O – 0.08%, N – 0.01%) is described by the transitional between parabolic and linear laws kinetics (<750 °C) or approximately linear kinetics (>750 °C), oxide scales consist of γ-ZrO_{2-x} with noticeable amount of elemental C in it (<750 °C) or α-ZrO_{2-x} (>750 °C); at $T <$ 750 °C oxide scales on carbide are protective, max. content of elemental C in the scales coincides with the beginning of their destruction as a result of the disturbance produced by CO release	
	O_2	~500–900	Both ZrC_{1-x} powders and single crystals start to interact with O_2 in the accordance with the following reaction: $ZrC_{1-x} + (y/2 + z)O_2 = ZrC_{1-x-z}O_y + zCO_2\uparrow$, as the process proceeds the oxycarbide-gas interface becomes saturated in O, forming the oxide scale: $ZrC_{1-x-z}O_y + (2 - y - \delta)/2\ O_2 = ZrO_{2-\delta} + (1 - x - z)C$, this layer is dense and protective, which results in parabolic kinetics, and contain some residual elemental C in amorphous state; once this layer reaches 1–10 μm in thickness, the stresses in it leads to crack formation, which results in linear kinetics, with the dense layer maintaining a constant thickness as new scale layer growth occurs, while the outer portions continue to crack forming pathways for O ingress to oxidize C in the oxide scale (the evolution of C oxides also contributes to additional crack and pore formation in the outer (non-protective) layer	

(continued)

Table 5.26 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	Reference
	Air	500–1000	The oxidation of hot-pressed $ZrC_{\sim 1.0}$ leads to the formation of intermediate Zr_2O (?), which almost completely transferred to ZrO_2 phases at higher temperatures	
	Air, static conditions	500–1200	At earlier stages (5 min exposure) oxidation of hot-pressed $ZrC_{\sim 1.0}$ (porosity $\leq$1%) results in the formation of the oxide scale, which consists mainly of γ-ZrO_{2-x} with C inclusions (at $\leq$900 °C metallic Zr and at $\geq$ 1000 °C α-ZrO_{2-x} were observed too); after 6 h exposure the scale composition is shifted to α/γ-ZrO_{2-x} – $ZrN_{1\pm x}$ mixture accompanied with a well-defined C layer (at $\leq$ 800 °C metallic Zr was observed too), the max. mass gain (up to 100 mg cm^{-2}) is observed at 800 °C[f]	
	O_2, 0.13–27 kPa	500–1200	At earlier oxidation stages (5–10 min exposure) of sintered $ZrC_{0.99}$ (porosity – 30%, contents: non-combined C – 0.20%, O $\leq$ 0.05%, N $\leq$ 0.05%) the effect of temperature is estimated by $E \approx 170$ kJ mol^{-1} and effect of oxygen pressure – by power (exponent) index $m = 1/3$	
	O_2 (0.02–2.0 kPa), Ar	500–1500	The isothermal oxidation (1–20 h exposure) of single crystal $ZrC_{0.97}$ (200), (220) leads to the formation of the oxide scale divided into outer layer (cracked and porous, growing linearly and formed by preferentially oriented α-ZrO_{2-x} and β-ZrO_{2-x} crystals containing 5–10 at.% C) and inner film (almost poreless, growing parabolically to finally attain a constant thickness and formed by amorphous (at <1100 °C) or relatively ordered (at >1100 °C) elemental (non-combined) C (up to 25 at.%) with dispersed γ-ZrO_{2-x} and α/β-ZrO_{2-x} particles in it, all of them – 1–10 nm in size); a considerable amount of C concentrated at the interface between carbide and inner film and gradient of O concentration both in carbide and inner film are observed, γ-ZrO_{2-x} crystallites are formed directly at the interface on the ZrC_{1-x} lattice while maintaining the above prefered orientation	
	O_2, 0.9–2.7 kPa	550–600	Oxidation of ZrC_{1-x} ($0.13 \leq x \leq 0.34$) phases is accompanied with the precipitation of elemental carbon, which is more intensive for the phases with less carbon deficit	

(continued)

Table 5.26 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	Reference
	O_2, 0.01–0.1 MPa	550–650	Oxidation kinetics of powdered ZrC_{1-x} is described by linear law with activation energy $E = 70 \pm 7$ kJ mol^{-1}; porous oxide scales formed on carbide are not protective	
	O_2, 0.13–27 kPa	550–1200	At earlier oxidation stages (5 min exposure) of sintered $ZrC_{0.66}$ (porosity – 5%, contents: non-combined C – 0.1%, O $\leq$ 0.05%, N $\leq$ 0.05%) the effect of temperature is estimated by $E = 160$ kJ mol^{-1} and effect of oxygen pressure – by power (exponent) index $m = ¼$	
	O_2, 0.13–27 kPa	550–1200	Oxidation kinetics of arc-melted $ZrC_{\sim 1.0}$ (content non-combined C – 5.0%) is described by parabolic law, the effect of temperature is estimated by $E \approx 135$ kJ mol^{-1} and effect of oxygen pressure – by power (exponent) index $m = ⅓$ (at 550–700 °C) and $m = 1/10$ (at 800–1200 °C); at $T \leq$ 1000 °C formed oxide scales are adherent to substrate only at earlier stages of oxidation	
	Air	575	A DTA-peak indicates onset of oxidation of Zr in powdered $ZrC_{0.98}$ (mean particle size – 10–40 μm, contents: non-combined C – 0.30%, O – 0.08%, N – 0.01%)	
	O_2 (0.5 kPa), Ar	580–900	The oxidation rate of powdered ZrC_{1-x} is proportional to p_{O2}, but virtually independent of T, as oxidation kinetics is controlled by the rate of O arrival at the reaction surface; γ-ZrO_{2-x} is the major oxidation product, though α-ZrO_{2-x} is observed too	
	O_2 (2.6 kPa), Ar	600	The oxidation of single crystal $ZrC_{0.97}$ (100) occurs through the dissolution of O accompanied with the precipitation of C: $ZrC_{1-x} + y/2\ O_2 = ZrC_{1-x-z}O_y + zC$, and subsequent formation of oxygen-deficit γ-ZrO_{2-x} oxide scale: $ZrC_{1-x-z}O_y + (2 - y - \delta)/2\ O_2 = ZrO_{2-\delta} + (1 - x - z)C$; high C concentration (up to 25 at.%) is observed at the carbide-oxide interface and in the oxide scale after 1 h exposure	
	O_2/Ar, ~20/80 kPa	600–700	Pure (single-phase) ZrC_{1-x} ceramics is oxidized completely during continuous heating with 20 °C min^{-1} rate	
	Air, static conditions	600–1000	At earlier stages (4–30 min exposure) of oxidation of arc-melted $ZrC_{0.61-0.83}$ materials the dissolution of O_2 and N_2 and formation of $ZrC_{1-x}N_yO_z$ oxycarbonitride phase is observed; the oxidation process of ZrC_{1-x} phases with greater deviation from the stoichiometry starts earlier	

(continued)

Table 5.26 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	Reference
	Air	600–1200	The rate of oxidation of arc-melted $ZrC_{0.61–0.96}$ materials (content non-combined C – 0.12–0.20%) increases with a decrease in the carbon content of monocarbide phases, the formed oxide scales consist of γ-ZrO_{2-x} and α-ZrO_{2-x} phases at all the temperatures, on the boundary of the scales and carbide phases an elemental C layer is formed by the reaction $ZrC_{1-x} + (2 - \delta)/2\ O_2 = ZrO_{2-\delta} + (1 - x)C$ followed by the formation of C oxides	
	Air (flow)	600–1200	Oxidation kinetics of powdered ZrC_{1-x} (mean grain size – 1.5 μm, specific surface area – 0.59 m^{-2} g) is ⅔-ordered (plots of the mass of remaining carbide, fraction unoxidized, is linear with time); the oxidation leads to the formation of γ-ZrO_{2-x} and α-ZrO_{2-x} with increasing amounts of α-ZrO_{2-x} at higher temperatures, and appearance of β-ZrO_{2-x} at $T \geq 1200$ °C	
	Air	650–700	The oxidation of CVD-ZrC_{1-x} coating leads to the formation of α-ZrO_{2-x} (mainly) and β-ZrO_{2-x} phases	
	Flow O_2/Ar, 0.8 Pa/ 0.16 MPa	675	During the non-isothermal oxidation of sintered $ZrC_{\sim 1.0}$ (porosity ~30%) with heating rate 10 K min^{-1}, the current reactivity of ~7 × 10^{-3} nmol cm^{-2} s^{-1} for carbide interaction with O_2 was detected	
	Air	700–800	Both γ-ZrO_{2-x} and α-ZrO_{2-x} phases are detected in the oxide scale formed on powdered $ZrC_{0.98}$ (mean particle size – 10–40 μm, contents: non-combined C – 0.30%, O – 0.08%, N – 0.01%); the attainment of max. content of non-combined C in the oxide scale coincides with the beginning of its destruction	
	O_2, 1.3–2.7 kPa	700–1000	Oxidation kinetics of sintered $ZrC_{0.92}$ (porosity $\leq 5\%$) is described by linear law, the effect of temperature is described by $E \approx 60$ kJ mol^{-1} and effect of oxygen pressure – by power (exponent) index $m = 1.3$ (oxidation process is controlled by a surface reaction at the α-ZrO_{2-x}-$ZrC_{1-x}O_y$ phase boundary)	
	O_2 (0.08–80 kPa), Ar	700–1500	The isothermal oxidation of single crystal $ZrC_{0.97}$ (200), (220) is described by linear kinetics, the duplex oxide scale consists of outer zone (cracked and porous, 6–11 at.% of elemental C) growing linearly and inner zone (poreless, 20–25 at.% of elemental C) growing parabolically	

(continued)

Table 5.26 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	Reference
	Air	725	A DTA-peak indicates onset of oxidation of C in powdered $ZrC_{0.98}$ (mean particle size – 10–40 μm, contents: non-combined C – 0.30%, O – 0.08%, N – 0.01%)	
	Air flow, ~5 cm s^{-1}	800	Oxidation mass gain of hot-pressed ZrC_{1-x} (porosity ~5%) is 20–50 mg cm^{-2} (exposure – 0.5–5.0 h)	
	Air	800–900	CVD-ZrC_{1-x} coating is oxidized completely; the oxide scale consists of α-ZrO_{2-x} only phase (1 h exposure)	
	Air, static conditions	800–1200	Recession rates (determined from the change in dimensions after the removal of oxide scale for 15 min exposure) of hot-pressed $ZrC_{\sim 1.0}$ are 28, 97 and 114 nm s^{-1} at 800, 1000 and 1200 °C, respectively ($E \approx 70$ kJ mol^{-1})	
	Air, O_2	800–1200	The "Maltese cross" oxidation mechanism of dense ZrC_{1-x} cuboids comprised 3 steps: delamination of sample edges, crack formation at corners and crack propagation towards the inner core with formation of microcracks parallel to the interface that increase the accessible surface area followed by a drastic volume expansion; the reaction interface comprises an amorphous C layer (thickness <2 μm) with α-ZrO_{2-x} nanocrystals (<5 nm) embedded in it	
	O_2/Ar, 10–13 Pa/ 0.1 MPa	810–820	The oxidation of single crystal $ZrC_{0.98}$ (content O < 0.002%) leads to the formation of γ-ZrO_{2-x} and α-ZrO_{2-x} phases containing elemental C (the latter one is in the amorphous state); during the cooling treatment, due to crack formation attributed to thermal expansion mismatch between carbide and oxides, a relatively small fraction of C is burnt with CO and CO_2 formation	
	O_2 (0.3–44 kPa), He (flow rate ~1 cm^3 s^{-1})	850–1900	Oxidation kinetics of zone-refined $ZrC_{0.96}$ (contents: O – 0.005%, N – 0.067%) is described by linear law with non-preferential interaction (i.e. Zr is oxidized at the same rate as C), the oxidation process is controlled by the rate of O_2 arrival at the surface, only α-ZrO_{2-x} phase was detected in oxide scales, while both CO and CO_2 were present in gaseous reaction products (materials disintegration due to the growth of oxide in pre-existing cracks and grain boundaries is observed at T = 850–1300 °C, the average oxidation mass gain is ~25–30 mg cm^{-2} for 2 h exposure at T = 1600–1900 °C and p_{O2} = 1.2 kPa)	

(continued)

Table 5.26 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	Reference
	O_2	~870	The penetration of O into the bulk ZrC_{1-x} materials is 175 μm (100 h exposure)	
	Air	900	The oxidation mass gain of ZrC_{1-x} hot-pressed materials is 25–46 mg cm^{-2} (exposure – 1–5 h)	
	Air	900–1000	Hot-pressed ZrC_{1-x} materials are oxidized considerably	
	O_2, 0.03 kPa	900–1300	The oxidation mass gain of sintered $ZrC_{0.92}$ (1–5% porosity, grain size – 50–100 μm) is 12–13 mg cm^{-2} (20 min exposure)	
	Air	980	Thickness of oxide scale on ZrC_{1-x} materials: 87.5 μm (50 h exposure) and 175 μm (100 h exposure)	
	Air	1000	The oxidation mass gain of ZrC_{1-x} hot-pressed materials is ~20 mg cm^{-2} (5 h exposure)	
	O_2, 4×10^{-3}–4×10^{-7} Pa	1000–1200	Surface of ZrC_{1-x} materials is covered by a layer of O atoms adsorbed on a $ZrC_{1-x}O_y$ monolayer	
	Air	1100–1200	The oxidation mass gain of ZrC_{1-x} hot-pressed materials is ~55–60 mg cm^{-2} (5 h exposure)	
	O_2, 3 Pa–2.7 kPa	1100–1500	Oxidation kinetics of sintered $ZrC_{0.92}$ (porosity $\leq$5%) is described by parabolic law °C and effect of oxygen pressure – by power (exponent) index $m = 0.3$ (1200 °C), "Maltese cross" dense oxide scales consist of inner (γ-ZrO_{2-x} ($x \approx 0.14$ at 1200 °C) stabilized by dissolved C (or ZrO_2C_x) and dispersed C) and outer (α-ZrO_{2-x}) layers; only ~70% of C remove from solids to gases to form CO_2, CO is observed in gaseous products only at $p_{O2} < 13$ Pa (oxidation process is controlled by diffusion of O to the γ-ZrO_{2-x}-$ZrC_{1-x}O_y$ phase boundary)	
	O_2, ~10^{-5} Pa	1200–1800	At earlier stages the oxidation of ZrC_{1-x} phases is accompanied with release of CO (for $x \approx 0$) or a CO_2 + CO mixture (for $x \gg 0$) in accordance with the following reactions: $ZrC_{1-x} + y/2\ O_2 = ZrC_{1-x-y} + yCO\uparrow$ (for $x \approx 0$, accompanied with mass loss), $ZrC_{1-x} + [(2 - z) + (1 - x)(1 + m)]/2\ O_2 = ZrO_{2-z} + m(1 - x)CO_2\uparrow + (1 - m)(1 - x)CO\uparrow$ (for $x \gg 0$, $m \gg$ ½, accompanied with mass gain)	

(continued)

Table 5.26 (continued)

System	Atmosphere	Temperature range, °C	Interaction character, products and/or compatibility	Reference
	Air flow, 41.5 m s^{-1} (0.72 d m^3 s^{-1})	1600–2200	Average recession rates (determined from the change in dimensions for 5 min exposure) of hot-pressed $ZrC_{\sim 1.0}$ are 0.5, 1.2 and 2.2 μm s^{-1} at 1600, 2000 and 2200 °C, respectively ($E \approx 70$ kJ mol^{-1})	
	O_2/C_2H_2 torch	~2000–3000 (heat fluxes 0.24–0.42 kW cm^{-2})	The CVD $ZrC_{\sim 1.0}$ coating (thickness ~75–85 μm) is oxidized forming transitional $ZrC_{1-x}O_y$ (thickness ~8–25 μm) and outer $ZrC_{\sim 2.0}$ layers, which act as a good O diffusion barrier and exhibit efficacious mechanical denudation resistance	
	O_2/C_2H_2 torch	≤~2800	The supersonic plasma-sprayed ZrC_{1-x} coating (thickness – 200 μm, contaminated with Zr oxides) has linear ablating rate – 0.9 μm s^{-1} and mass gain rate – 2.0 mg s^{-1} (30 s exposure); during the ablation process the reaction $ZrC_{1-x} + [2 + (1 + m)(1 - x)]/2\ O_2 = ZrO_2 + + m(1 - x)CO_2\uparrow + (1 - m)(1 - x)CO\uparrow$ produce the liquid oxide layer, which acts as a thermal barrier and reduces the diffusion of O into the bulk material	
	O_2/C_2H_2 torch	≤~2800	The CVD $ZrC_{0.70}$ coating (thickness > 200 μm) has ablating rates: linear – 0.3 μm s^{-1} and mass – 0.1 mg cm^2 s^{-1} (4 min exposure)	
	O_2 (~1 Pa), Ar	≥3300 (laser heating, ~25 kW cm^{-2})	The surface heat treatment of $ZrC_{0.82}O_{0.14}$ leads to the evolution of O in the form of the gaseous species ZrO and conversion of oxycarbide into the pure carbide phase of $ZrC_{0.75}$, while the grains of solids grow considerably; the traces of O_2 present in the treatment cell react with the carbide surface also giving ZrO gas, and do not form any oxidised solid phases (oxide scales) *See also* section C–O–Zr in Table I-2.14	
ZrC_{1-x}–Xe			*See* Table 5.27	

[a]Under the conditions simulated rocket combustion gases at the surface temperatures of ~2250 °C sintered ZrC_{1-x} was oxidized according to a parabolic law owing to the formation of an adherent oxide film (scale) [932]

[b]In this temperature range the chlorination can be also carried out in molten salts (e.g. LiCl-KCl melts) by means of electrochemical procedures (both dissolved chlorine and chlorine gas bubbles react with ZrC_{1-x}

[c]The analogous Ti_2CH and Hf_2CH phases were synthesized via hydrogen dissolution in non-stoichiometric carbides (dehydrogenation leads to the starting carbides again) by Goretzki et al. [1142, 1143]

[d]Fine zirconium carbide powders are pyrophoric and susceptible to oxidation at room temperature [39, 46, 96]

[e]For near-stoichiometric ZrC_{1-x} the value of the Pilling-Bedworth ratio $\alpha = M_O d_C / M_C d_O = 1.43$, where M_O is molecular mass of the oxide phase formed on the oxidation of 1 mol of carbide phase, M_C is molecular mass of carbide phase, d_C and d_O are the densities of carbide and oxide phases, respectively [1253]

[f]Termed as a ridge temperature in accordance with Shabalin's "ridge effect" model [984, 985]

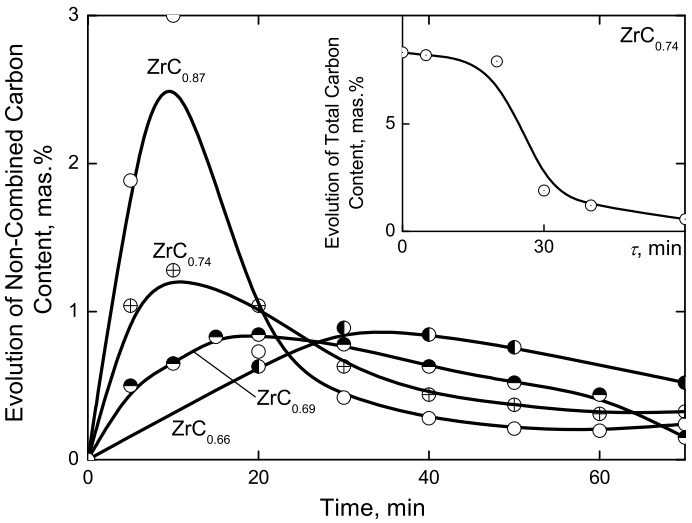

Fig. 5.37 Evolution of non-combined carbon content during the isobaric-isothermal oxidation of flat-plate samples of sintered ZrC_{1-x} materials with various deviations from the stoichiometry (porosity in the range of 5–15%) in the gas flow of oxygen with pressure of 2.67 kPa at temperature of 600 °C (*Inset* – evolution of total carbon content (corresponding to the total burning-off of carbon) in $ZrC_{0.74}$ during the oxidation process at the same conditions) [983]

ZrC_{1-x} phases are given in Table 5.27. Data on the catalytic activities of zirconium carbide are given in several works [1, 515, 710, 1089, 1094–1122, 1384, 1461, 1502]; the electrochemical behaviour of zirconium monocarbide materials in various media is described elsewhere [933, 1120–1130, 1345, 1346]. The parameters of wettability of ZrC_{1-x} phases with some liquid metals (melts) are listed in Table 5.28, the diffusion rates for the systems containing zirconium carbide phases are presented in species pairs within the various ranges of temperatures in the Table 5.29. The characters of chemical interaction of ZrC_{1-x} and $ZrC_{1-x}O_y$ phases with some common chemicals (acids, alkalies and salts in aqueous solutions and melts) are summarized in Tables 5.30 and 5.31.

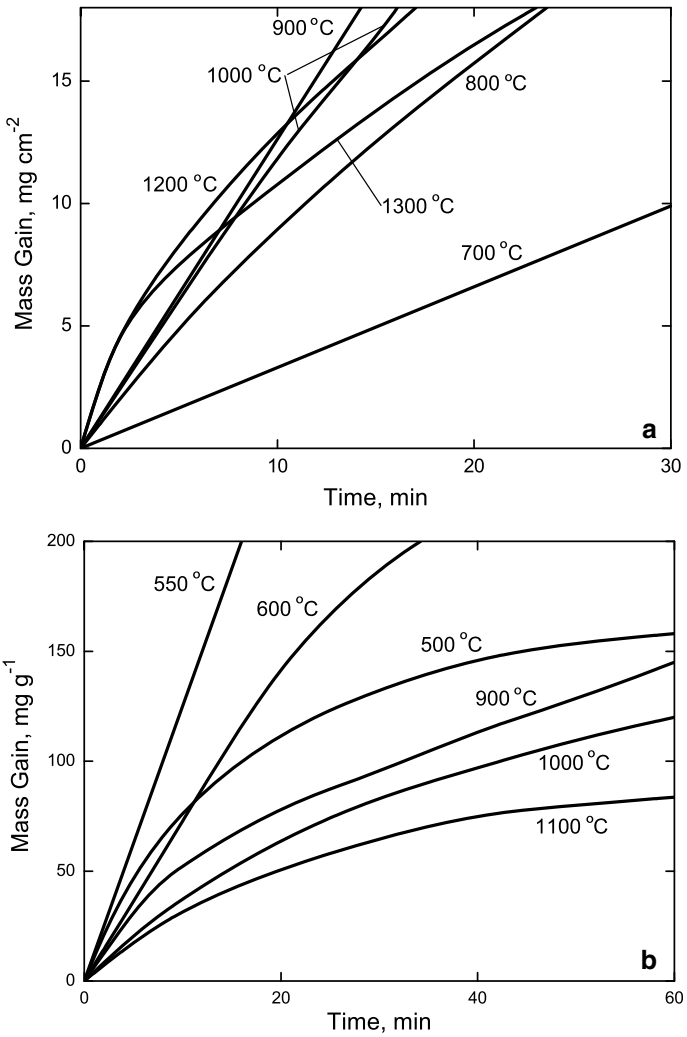

Fig. 5.38 Isothermal-isobaric oxidation kinetics curves for the flat-plate samples of sintered ZrC_{1-x} materials in gas flow at lower oxygen pressure p_{O2}: (**a**) $ZrC_{0.92}$ (open porosity $\leq$ 5%), p_{O2} = 1.33 kPa [980, 981]; (**b**) $ZrC_{0.99}$ (porosity – 30%, content non-combined C – 0.2%), p_{O2} = 0.40 kPa [982]

Table 5.27 The radiochemical compatibility of ZrC_{1-x} materials with some fission products (in alphabetical order) and diffusion parameters of these products in carbide phases at irradiation and/or annealing temperatures

Fission product	Carbide phase composition	Temperature range, °C	Neutron irradiation fluence, cm^{-2}	Diffusion coefficient, $cm^2\ s^{-1}$	Activation energy kJ mol^{-1}	Release %	Reference
^{110m}Ag	$ZrC_{\sim 1.0}$	900[a]/1500(10^4 h)[b]	3.4×10^{25} (fast)	–	–	–[c]	[96, 1211]
	$ZrC_{\sim 1.0}$	1200[a]/1500(10^4 h)[b]	6.5×10^{25} (fast)	–	–	–	[96, 1211]
	$ZrC_{\sim 1.0}$	~1200–1700	–	–	~80[d] and ~225[e]	–	[96, 1225]
	$ZrC_{\sim 1.0}$	900[a]	1.2×10^{25} (fast)	–	–	Good retention	[770, 773]
^{133}Ba	$ZrC_{\sim 1.0}$	~1200–1700	–	–	~290	–	[96, 1225]
^{140}Ba	$ZrC_{\sim 1.0}$	1400(220 h)	–	10^{-13}–10^{-14}	–	Good retention	[96, 1227]
^{144}Ce	$ZrC_{\sim 1.0}$ (content non-combined C – 3.4%)	1260(1040 h)[a]/1600[b]	1.4×10^{21} (fast)	~2.0×10^{-12}	–	<0.01[f]	[4, 96, 781]
	$ZrC_{\sim 1.0}$	900[a]/1500(10^4 h)[b]	3.4×10^{25} (fast)	–	–	Good retention	[96, 1211]
	$ZrC_{\sim 1.0}$	1200[a]/1500(10^4 h)[b]	6.5×10^{25} (fast)	–	–	Good retention	[96, 1211]
	$ZrC_{\sim 1.0}$	1200[a]	5.0×10^{21} (fast)	–	–	Good retention	[96, 767]
	$ZrC_{\sim 1.0}$	–	–	–	–	Good retention	[96, 1229]
^{134}Cs/^{137}Cs	$ZrC_{\sim 1.0}$	900[a]	1.2×10^{25} (fast)	–	–	2[g]	[770, 773]
	$ZrC_{\sim 1.0}$	1600(4500 h)[b]	1.2×10^{25} (fast)	$(1–5) \times 10^{-14}$	–	0.03	[96, 770, 773]
	$ZrC_{\sim 1.0}$	1800(3000 h)[b]	1.2×10^{25} (fast)	1×10^{-13}–2×10^{-14}	–	0.04	[96, 770, 772]
	$ZrC_{\sim 1.0}$	2000(100 h)[b]	1.2×10^{25} (fast)	–	–	10	[770, 772]
^{137}Cs	$ZrC_{\sim 1.0}$	1200[a]	5.0×10^{21} (fast)	–	–	–[c]	[96, 767]
	$ZrC_{0.98}$	~1200–1600	–	1.4×10^{-12}–3.1×10^{-14}	~50	–	[96, 867]
	$ZrC_{\sim 1.0}$ (content non-combined C – 3.4%)	1260(1040 h)[a]/1500–1600[b]	1.4×10^{21} (fast)	6.0×10^{-14} /4.0×10^{-12}–6.0×10^{-13}, or $D = 2.3 \times 10^{-3} \times \exp(-38{,}200/T)$ (*see also* Table 5.29)	~310–320	–	[4, 96, 781]
	$ZrC_{\sim 1.0}$	900[a]/1500(10^4 h)[b]	3.4×10^{25}	–	–	Good retention	[96, 1211]
	$ZrC_{\sim 1.0}$	1200[a]/1500(10^4 h)[b]	6.5×10^{25}	–	–	Good retention	[96, 1211]

(continued)

Table 5.27 (continued)

Fission product	Carbide phase composition	Temperature range, °C	Neutron irradiation fluence, cm^{-2}	Diffusion coefficient, $cm^2\ s^{-1}$	Activation energy $kJ\ mol^{-1}$	Release %	Reference
^{154}Eu	$ZrC_{\sim 1.0}$	900[a]/1500(10^4 h)[b]	3.4×10^{25}	–	–	Good retention	[96, 1211]
	$ZrC_{\sim 1.0}$	1200[a]/1500(10^4 h)[b]	6.5×10^{25}	–	–	Good retention	[96, 1211]
	$ZrC_{\sim 1.0}$	1600[b]	1.2×10^{25} (fast)	–	–	>0.03 (4500 h)?	[96, 773]
^{155}Eu	$ZrC_{\sim 1.0}$	1600[b]	1.2×10^{25} (fast)	–	–	>0.03 (4500 h)?	[96, 773]
$^{(3,4)}He$	$ZrC_{0.92\pm0.0}$	1000–1600[h]	–	–	74 ± 8 (*see also* Table 5.29)	–	[1360]
	$ZrC_{\sim 1.0}$	−70–1730	–	$5.13 \times 10^{-7}\exp(-9{,}050/T)$ (*see also* Table 5.29)[i]	75	–	[808, 1361]
^{88}Kr	$ZrC_{\sim 1.0}$	900[a]	1.2×10^{25} (fast)	–	–	$(2–4) \times 10^{-5}$	[768, 770]
^{95}Nb	$ZrC_{\sim 1.0}$	1400	–	10^{-14}	–	–	[96, 1227]
$^{(107)}Pd$	$ZrC_{\sim 1.0}$	1550–1850[h]	–	–	~300[j]	–	[96, 1014]
	$ZrC_{\sim 1.0}$	1200–1500[h]	–	–	~210–270[j]	–	[1284]
^{103}Ru	$ZrC_{\sim 1.0}$	1400	–	10^{-12}	–	–	[96, 1227]
^{106}Ru	$ZrC_{\sim 1.0}$	1600(500 h)[b]	1.2×10^{25} (fast)	3×10^{-12}	–	0.4	[96, 770]
	$ZrC_{\sim 1.0}$	1600(4500 h)[b]	1.2×10^{25} (fast)	–	–	13	[96, 770, 773]
	$ZrC_{\sim 1.0}$	1800(3000 h)[b]	1.2×10^{25} (fast)	–	–	86	[96, 770, 773]
^{85}Sr	$ZrC_{\sim 1.0}$	1400	–	$\sim 2 \times 10^{-14}$	–	–	[96, 1227]
^{90}Sr	$ZrC_{\sim 1.0}$ (content non-combined C – 3.4%)	1260(1040 h)[a]/1600[b]	1.4×10^{21} (fast)	$\sim 1.5 \times 10^{-11}$	–	–	[4, 96, 781]

(continued)

Table 5.27 (continued)

Fission product	Carbide phase composition	Temperature range, °C	Neutron irradiation fluence, cm^{-2}	Diffusion coefficient, $cm^2\ s^{-1}$	Activation energy $kJ\ mol^{-1}$	Release %	Reference
^{133}Xe ($\sim 1.0 \times 10^{-7}$ at%)	$ZrC_{0.58–0.64}$ (content O – 0.67%)	~1200–1650[b]	$1.0 \times 10^{13}\ cm^{-2}\ s^{-1}$ (thermal)[k]	10^{-4}–10^{-6}	~240–330	75 (465 h)	[1228]
	$ZrC_{0.58–0.64}$ (content O – 0.67%)	~1200–1800[b]	$1.0 \times 10^{13}\ cm^{-2}\ s^{-1}$ (thermal)[k]	–	–	84 (790 h)	[1228]
	$ZrC_{0.76–0.79}$ (content O – 0.64%)	~1200–1800[b]	$1.0 \times 10^{13}\ cm^{-2}\ s^{-1}$ (thermal)[k]	10^{3}–10^{4}	~660	30 (700 h)	[1228]
	$ZrC_{0.83–0.85}$ (content O – 0.31%)	~1250–1850[b]	$1.0 \times 10^{13}\ cm^{-2}\ s^{-1}$ (thermal)[k]	10^{5}–10^{6}	~790	30 (760 h)	[1228]
	$ZrC_{0.91–0.92}$ (content O – 0.04%)	~1300–1900[b]	$1.0 \times 10^{13}\ cm^{-2}\ s^{-1}$ (thermal)[k]	10^{3}–10^{4}	~770	23 (770 h)	[1228]
	$ZrC_{0.91–0.92}$ (content O – 0.04%)	~1200–1900[b]	$1.0 \times 10^{13}\ cm^{-2}\ s^{-1}$ (thermal)[k]	–	–	30 (570 h)	[1228]
	$ZrC_{\sim 1.0}$ (content O – 0.05%)	1200–1500[b]	$1.0 \times 10^{13}\ cm^{-2}\ s^{-1}$ (thermal)[k]	10^{-6}	~340	49 (650 h)[l]	[1228]
	$ZrC_{\sim 1.0}$ (content O – 0.05%)	1500–1700[b]	$1.0 \times 10^{13}\ cm^{-2}\ s^{-1}$ (thermal)[k]	5.0×10^{-3}	~460	49 (650 h)[l]	[1228]
	$ZrC_{\sim 1.0}$ (content O – 0.05%)	1650–1850[b]	$1.0 \times 10^{13}\ cm^{-2}\ s^{-1}$ (thermal)[k]	10	~610	49 (650 h)[l]	[1228]
^{136}Xe	$ZrC_{0.80–0.95}O_{0.05–0.20}$	1600–1800[h]	–	–	–	Xe migration is enhanced in O-rich phases	[1322]

[a]Irradiation temperature
[b]Post-irradiation heat treatment temperature
[c]Release is below the detectable limit
[d]For migration along the short circuit paths
[e]For migration in the grain volumes
[f]Mostly through grain boundaries with some contribution from volume diffusion
[g]The value compared to the concentration in the kernel
[h]Annealing temperature
[i]Obtained from first-principles total energy and phonon frequency calculations on the basis of the transition state theory (TST)
[j]The value of the activation energy of $ZrPd_3$ formation
[k]The value of thermal neutron flux
[l]Summarized for the temperature range from 1200 to 1850 °C

Table 5.28 The parameters of wettability of zirconium monocarbide ZrC_{1-x} phases with some liquid metals and alloys (melts)[a]

Melt (purity)	Atmosphere	Temperature, °C	Time, s	$\gamma_{l\text{-}g}$, mJ m^{-2}	W_a, mJ m^{-2}	W_m^b, kJ mol^{-1}	θ, degree	Reference
				$ZrC_{0.60}$				
Cu	Vacuum	1100	1200	1357	2385	–	~40	[1, 32, 1240]
				$ZrC_{0.72}$				
Cu	Vacuum	1100	1200	1351	1585	–	~80	[1, 32, 1240]
				$ZrC_{0.78}$				
Cu	Ar	1130	900	1351	270	–	143 ± 2	[1, 1231]
				$ZrC_{0.86}$				
Cu	–	~1100	–	–	–	–	141	[1241, 1244]
				$ZrC_{0.88}$				
Cu	Vacuum	1100	900	1352	780	–	~115	[1, 32, 1240]
Cu	Ar	1130	900	1351	365	–	137 ± 1	[1, 1231]
				$ZrC_{0.90}$				
Cu	Ar	1130	900	1351	332	–	139 ± 1	[1, 1231]
Cu	–	~1100	–	–	–	–	141	[1241, 1244]
				$ZrC_{0.92}$				
Cu	–	~1100	–	–	–	–	142	[1241, 1244]
Cu[c]	Vacuum, 1.3 mPa	1150	3600	1000	–	–	128	[1527]
Cu (Zr – 1.4%)[c]	Vacuum, 1.3 mPa	1150	3600	–	–	–	138	[1527]
Cu (Zr – 1.4, V – 0.8%)[c]	Vacuum, 1.3 mPa	1150	3600	–	–	–	112.5	[1527]
Ni[d]	Vacuum	1450	1500	–	–	–	25	[1351]
Ni (C-0.7%)[d,e]	Vacuum	1450	1500	–	3100	159	35	[1, 1351, 1522]
Ni (C-1.0%)[d]	Vacuum	1450	1500	–	–	–	60	[1, 1351, 1532]
Ni (C-1.4%)[d,f]	Vacuum	1450	1500	–	2300	117	68	[1, 1351, 1522]
Ni (C-1.8%)[d]	Vacuum	1450	1500	–	–	–	70	[1, 1351, 1522]
Ni (C-2.8%)[d,g]	Vacuum	1450	1500	–	2000	105	80	[1, 1351, 1522]
Ni (C-3.5%)[d,h]	Vacuum	1450	1500	–	1700	88	91	[1, 1351, 1522]
Ni (C-7.2%)[d,i]	Vacuum	1450	1500	–	1300	67	103	[1, 1351, 1522]
				$ZrC_{0.96}$				
Cu	–	~1100	–	–	–	–	143	[1241, 1244]

(continued)

Table 5.28 (continued)

Melt (purity)	Atmosphere	Temperature, °C	Time, s	$\gamma_{l\text{-}g}$, mJ m^{-2}	W_a, mJ m^{-2}	W_m^b, kJ mol^{-1}	θ, degree	Reference
				$ZrC_{0.97}$				
Co[j]	Vacuum	1500	900	1805	3260	152	36	[1, 851]
Cu	Ar	1130	900	1351	332	–	139 ± 2	[1, 1231]
Fe[k]	Vacuum	1550	900	1780	2940	155	49	[1, 851]
Fe (C-4.0, Mn-1.4, Si-2.5, S-0.1%)[l]	Pure Ar	1450	900	–	–	–	125	[1, 48, 1242]
	Pure Ar	1450	1200	–	–	–	111	[1, 1242]
	Pure Ar	1450	2400	–	–	–	95	[1, 1242]
	Pure Ar	1450	3600	–	–	–	85	[1, 1242]
	Pure Ar	1450	7200	–	–	–	83	[1, 1242]
Fe (C-3.1, Mn-0.4, Si-1.6, S-0.02%)[l]	Pure Ar	1400–1450	900	–	–	–	100	[1, 48, 1242]
Fe (C-2.7, Mn-0.4, Si-1.2, S-0.2%)[l]	Pure Ar	1400–1450	900	–	–	–	122	[1, 48, 1242]
Fe (C-2.6, Mn-1.0, Si-1.0, Cr-0.3%)[l]	Pure Ar	1400–1450	900	–	–	–	132	[1, 48, 1242]
Fe (C-1.2, Mn-0.2, Si-0.2, P-0.03, S-0.02%)[l]	Pure Ar	1550	900	–	–	–	110	[1, 1242]
Fe (C-1.0, Cr-1.4, Mn-0.4, Si-0.3%)[l]	Pure Ar	1500	900	–	–	–	122	[1, 48, 1242]
Fe (C-1.0, Cr-1.4, Mn-1.1, Si-0.6%)[l]	Pure Ar	1500–1550	900	–	–	–	145	[1, 48, 1242]
Fe (C-1.0, Mn-0.2, Si-0.2, P-0.03, S-0.02%)[l]	Pure Ar	1550	900	–	–	–	119	[1, 1242]
Fe (C-0.8, Mn-0.2, Si-0.2, S-0.03%)[l]	Pure Ar	1500	900	–	–	–	131	[1, 48, 1242]
Fe (C-0.5, Mn-0.6, Si-0.2, P-0.03, S-0.03%)[l]	Pure Ar	1500–1550	900	–	–	–	128	[1, 1242]
Fe (C-0.5, Mn-0.6, Si-0.2, P-0.03, S-0.035%)[l]	Pure Ar	1500–1550	900	–	–	–	135	[1, 1242]
Fe (C-0.5, Mn-0.6, Si-0.2, P-0.03, S-0.07%)[l]	Pure Ar	1500–1550	900	–	–	–	110	[1, 1242]

(continued)

Table 5.28 (continued)

Melt (purity)	Atmosphere	Temperature, °C	Time, s	$\gamma_{l\text{-}g}$, mJ m^{-2}	W_a, mJ m^{-2}	W_m^b, kJ mol^{-1}	θ, degree	Reference
Fe (C-0.5, Mn-0.6, Si-0.2, P-0.03, S-0.11%)[l]	Pure Ar	1500–1550	900	–	–	–	90	[1, 1242]
Fe (C-0.5, Mn-0.6, Si-0.2, P-0.03, S-0.13%)[l]	Pure Ar	1500–1550	900	–	–	–	76	[1, 1242]
Fe (C-0.2, Cr-14.5, Ni-2.7, Mn-0.5%)[l]	Pure Ar	1500–1550	900	–	–	–	102	[1, 48, 1242]
Fe (C-0.2, Mn-0.4, Si-0.06, S-0.03, P-0.03%)[l]	Pure Ar	1550	900	–	–	–	130	[1, 1242]
Ni[m]	Vacuum	1450	900	1700	3240	151	25	[1, 851]
				$ZrC_{\sim 1.0}$				
Ag	He	1100	1200	910	495	–	117 ± 2	[1, 1239]
Ag	He	1200	1200	910	510	–	116 ± 2	[1, 1239]
Ag	He	1300	1200	910	510	–	116 ± 2	[1, 1239]
Ag	He	1400	1200	910	705	–	103 ± 2	[1, 1239]
Ag	He	1500	1200	910	880	–	92 ± 2	[1, 1239]
Ag	He	1600	1200	910	1005	–	84 ± 2	[1, 1239]
Al (99.97%)	Vacuum	900–1000	900	914	122	6.5	150 ± 1	[1, 32, 48, 1231, 1232]
Al	Vacuum, Ar	1000	900	914	803	–	97	[1, 1233]
Al[n]	He	1000	900	914	430	–	122 ± 3	[1, 812]
Al[n]	He	1200	900	865	485	–	116 ± 3	[1, 812]
Al[n]	He	1400	900	865	1297	–	60 ± 3	[1, 812]
Au	Vacuum	–	1800	1140	143	–	151	[1, 32, 1230, 1240]
Bi (99.99%)	Vacuum	320–1200	900	390 ± 1	87	5	141 ± 1	[1, 32, 48, 1231, 1232]
Co (99.99%)[d]	Vacuum	1500	900	1805	3260	175	36	[1, 851, 1231, 1235]
Co (99.99%)	Ar	1500	900	1805	3550	186	15	[1, 32, 48, 1230, 1232]
Co	Vacuum, Ar	1495	–	–	–	–	86	[32, 48, 1230, 1233]
Co[o]	Vacuum	1420	300	1910	3460	–	36	[1, 32, 1230, 1234]
Cu (99.99%)	Vacuum	1130	900	1351	315	17	140 ± 1	[1, 48, 1231]
Cu (99.99%)	Ar	1370	900	1351	315	–	140	[1, 1230–1232]
Cu (99.99%)	Ar	1500	900	1351	395	–	135	[1, 1231, 1232]
Cu	Vacuum, Ar	1100	–	–	–	–	135	[31, 42, 1233, 1246]

(continued)

Table 5.28 (continued)

Melt (purity)	Atmosphere	Temperature, °C	Time, s	$\gamma_{l\text{-}g}$, mJ m^{-2}	W_a, mJ m^{-2}	W_m^b, kJ mol^{-1}	θ, degree	Reference
Cu	Ar	1100	900	1351	315	–	~140	[1, 31, 32, 48, 1238, 1240]
Cu	Ar	1500	900	1351	715	–	118	[1, 31, 32, 48, 1230, 1238]
Cu[p]	Vacuum	1100	900	1270	505	–	127	[1, 1234, 1528]
Cu[p]	Vacuum	1150	900	1255	525	–	126	[1, 1234]
Cu[p]	Vacuum	1200	900	1240	540	–	124	[1, 1234]
Cu (Ni-0.01%)	Vacuum, Ar	1200	1200	–	–	–	96	[1, 30, 31, 42, 1238, 1246]
Cu (Ni-0.05%)	Vacuum, Ar	1200	1200	–	–	–	70	[1, 30, 31, 42, 1238, 1246]
Cu (Ni-0.1%)	Vacuum, Ar	1200	1200	–	–	–	63	[1, 30, 31, 42, 1238, 1246]
Cu (Ni-0.25%)	Vacuum, Ar	1200	1200	–	–	–	54	[1, 30, 31, 42, 1246]
Fe (99.99%)[k]	Vacuum	1550	900	1780	2950	155	49	[1, 32, 851, 1230–1232]
Fe (99.99%)	Ar	1550	900	1780	420	32	140	[1, 32, 851, 1230–1232]
Fe	Vacuum, Ar	1535	–	–	–	–	50	[32, 48, 1230, 1233]
Fe[q]	Vacuum	1490	<60	1900	3245	–	45	[1, 32, 1230, 1234]
Fe (C-0.5%)[r]	Vacuum	1550	–	–	–	–	78	[1]
Fe (C-1.0%)[r]	Vacuum	1550	–	–	–	–	82	[1]
Fe (C-2.0%)[r]	Vacuum	1550	–	–	–	–	85	[1]
Fe (C-3.0%)[r]	Vacuum	1550	–	–	–	–	92	[1]
Fe (C-4.0%)[r]	Vacuum	1550	–	–	–	–	123	[1]
Fe (C-3.1–3.4, Si-1.7–2.1, Mn-0.8–1.2%)	Vacuum	1550	–	–	492	–	121	[1]
Fe (C-0.2, Mn-0.3, Si-0.2, W-0.07%)	Pure Ar	1550	–	–	–	–	~50	[1328]
	Pure Ar	1600	–	–	–	–	~40	[1328]
Fe (C-0.32–0.40, Mn-0.4–0.9, Si-0.20–0.52%)	Vacuum	1550	1200	–	1385	–	81	[1]
Fe–Cr[s] (Cu-3%)[t]	Vacuum	1200	900	–	–	–	42	[1, 1247]
Fe–Cr[s] (Cu-10%)[t]	Vacuum	1200	900	–	–	–	83	[1, 1247]
Fe–Cr[s] (Cu-25%)[t]	Vacuum	1200	900	–	–	–	83	[1, 1247]
Fe–Cr[s] (Cu-40%)[t]	Vacuum	1200	900	–	–	–	60	[1, 1247]

(continued)

Table 5.28 (continued)

Melt (purity)	Atmosphere	Temperature, °C	Time, s	$\gamma_{l\text{-}g}$, mJ m^{-2}	W_a, mJ m^{-2}	W_m^b, kJ mol^{-1}	θ, degree	Reference
Fe–Cr^s (Cu-60%)^t	Vacuum	1200	900	–	–	–	56	[1, 1247]
Fe–Cr^s (Cu-70%)^t	Vacuum	1200	900	–	–	–	56	[1, 1247]
Fe–Cr^s (Cu-75%)^t	Vacuum	1200	900	–	–	–	45	[1, 1247]
Fe–Cr^s (Cu-90%)^t	Vacuum	1200	900	–	–	–	61	[1, 1247]
Fe (Ni-2.5%)	Vacuum	1550	1200	–	–	–	39	[1, 1237]
Fe (Ni-5%)	Vacuum	1550	1200	–	–	–	30	[1, 1237]
Fe (Ni-25%)	Vacuum	1550	1200	–	–	–	21	[1, 1237]
Fe (Ni-30%)	Vacuum	1550	1200	–	–	–	27	[1, 1237]
Fe–Si^t	Vacuum	1200	900	–	–	–	84	[1, 1247]
Fe–Si^u (Cu-2%)^t	Vacuum	1200	900	–	–	–	45	[1, 1247]
Fe–Si^u (Cu-25%)^t	Vacuum	1200	900	–	–	–	90	[1, 1247]
Fe–Si^u (Cu-50%)^t	Vacuum	1200	900	–	–	–	40	[1, 1247]
Fe–Si^u (Cu-60%)^t	Vacuum	1200	900	–	–	–	62	[1, 1247]
Fe–Si^u (Cu-70%)^t	Vacuum	1200	900	–	–	–	52	[1, 1247]
Fe–Si^u (Cu-75%)^t	Vacuum	1200	900	–	–	–	38	[1, 1247]
Fe–Si^u (Cu-90%)^t	Vacuum	1200	900	–	–	–	66	[1, 1247]
Ga (99.999%)	Ar	1300	900	707	225	–	138	[1, 1231]
Ga (99.99%)	Vacuum, Ar	800	900	707	215	11	134 ± 1	[1, 32, 48, 1230, 1232]
Ge (99.999%)	Vacuum	1000	–	–	–	–	135	[32, 48, 1230]
Ge	Ar	1100	900	700	176	10	135 ± 1	[1, 1231, 1232]
Ge	Ar	1300	900	700	176	10	133 ± 1	[1, 1231, 1232]
In (99.99%)	Vacuum, Ar	250–700	900	559	113	6	143 ± 3	[1, 32, 48, 1230–1232]
Mn (99.8%)	Vacuum	1300	900	1750	2405	130	68	[1, 1231]
Mn (99.8%)	Ar	1300	900	1750	2200	116	75	[1, 1231, 1232]
Mn	Vacuum, Ar	1245	–	–	–	–	70	[32, 48, 1230, 1233]
Ni (99.99%)^m	Vacuum	1450	900	1700	3240	170	25	[1, 851, 1231]
Ni (99.99%)	Ar	1450	900	1700	3240	165	32	[1, 32, 48, 1230–1232]

(continued)

Table 5.28 (continued)

Melt (purity)	Atmosphere	Temperature, °C	Time, s	$\gamma_{l\text{-}g}$, mJ m^{-2}	W_a, mJ m^{-2}	W_m^b, kJ mol^{-1}	θ, degree	Reference
Ni (99.98%)[v]	Ar (99.999%)	1480	–	1700	–	–	73 ± 0.5	[1440]
Ni (99.98%)[v]	Ar (99.999%)	1480	400	1700	–	–	32 ± 0.8	[1440]
Ni	Vacuum	1380	–	–	–	–	24	[32, 1230, 1533]
Ni	Vacuum, Ar	1455	–	1700	–	–	43	[32, 48, 1230, 1233]
Ni[w]	Vacuum	1500	900	1700	3240	–	32	[1, 1232]
Ni[x]	Vacuum	1500	900	1700	3395	–	5	[1, 1232]
Ni	Vacuum	1550	1200	1700	3170	–	30	[1, 1237]
Ni[y]	Vacuum	1380	300	1810	3460	–	24	[1, 32, 1234, 1236]
Ni (P-0.1%)	Vacuum	1450	1200	–	–	–	15	[1]
Ni (P-1.5%)	Vacuum	1450	1200	–	–	–	6	[1]
Ni (P-2.5%)	Vacuum	1450	1200	–	–	–	4	[1]
Ni (P-4.8%)	Vacuum	1450	1200	–	–	–	2	[1]
Ni (P-7.2%)	Vacuum	1450	1200	–	–	–	33	[1]
Ni (P-12.0%)	Vacuum	1450	1200	–	–	–	82	[1]
Ni (P-20.0%)	Vacuum	1450	1200	–	–	–	95	[1]
Ni (Co-10, W-10, Cr-9.5, Al-5.5, Ti-2.5, Mo-2%)[z]	Vacuum, ~6.7 mPa	1450	60	–	–	–	110	[1]
	Vacuum, ~6.7 mPa	1450	300	–	–	–	60	[1]
	Vacuum, ~6.7 mPa	1450	600	–	–	–	52	[1]
	Vacuum, ~6.7 mPa	1450	1200	–	–	–	46	[1]
	Vacuum, ~6.7 mPa	1450	1800	–	–	–	46	[1]
Ni (W-9.5–11.0, Co-9.0–10.5, Cr-8.0–9.5%)[z]	He	1400	1200	–	2760	–	46	[1]
Ni (Cr-9.5–12.0, Al-5.0–6.0, W-4.5–5.5%)[z]	He	1350	1200	–	840	–	30	[1]
Ni (Co-10.0–15.0, Cr-8.5–10.5, Al-5.0–5.7%)[z]	He	1350	1200	–	1770	–	78	[1]
Ni (Cr-13.0–16.0, Mo-2.8–3.2, Al-1.3–1.7%)[z]	He	1380	1200	–	1660	–	90	[1]
Pb (99.99%)	Vacuum	400–1200	900	480	77	4	147 ± 1	[1, 32, 48, 1230–1232]

(continued)

Table 5.28 (continued)

Melt (purity)	Atmosphere	Temperature, °C	Time, s	$\gamma_{l\text{-}g}$, mJ m^{-2}	W_a, mJ m^{-2}	W_m^b, kJ mol^{-1}	θ, degree	Reference
Sb (99.999%)	Vacuum	700–920	900	384	260	13	110 ± 7	[1, 32, 48, 1230–1232]
Si (99.999%)	Vacuum, Ar	1500	900	860	1655	87.1	22 ± 1	[1, 32, 48, 1231, 1232]
Si	Vacuum, Ar	1410	–	–	–	–	57	[32, 48, 1233]
Sn (99.99%)	Vacuum, Ar	300–1200	900	554	74	4	150 ± 1	[1, 32, 48, 1230–1232]
Tl (99.99%)	Vacuum	400	900	490	188	10	128 ± 2	[1, 32, 48, 1230–1232]

[a]The parameters of wettability are given in accordance with Young-Dupré equation $W_a = \gamma_{l\text{-}g} \times (1 + \cos\theta)$ and Young's equation $\gamma_{s\text{-}l} = \gamma_{s\text{-}g} - \gamma_{l\text{-}g}\cos\theta$, where W_a is the work of adhesion, $\gamma_{l\text{-}g}$ is the liquid-vapour interfacial energy (surface tension), $\gamma_{s\text{-}l}$ is solid-liquid interfacial energy, $\gamma_{s\text{-}g}$ is the solid-vapour interfacial energy and θ is the wetting contact angle [1]; compositions of melts are given in mass (weight) percentage

[b]$W_m = W_a(M/d)^{2/3}N_A^{1/3}$, where W_m is the molar work of adhesion, M is the molecular mass and d is the density of chemical compound, N_A is the Avogadro constant [1351]

[c]Sintered materials (porosity – ≤2.0%, contents: non-combined C – 0.30%, O – 0.015%, N – 0.27%, Fe – 0.10%)

[d]Porosity ≤ 5%, content non-combined C – 0.3%

[e]$\gamma_{s\text{-}l}$ = 1300 mJ m^{-2}

[f]$\gamma_{s\text{-}l}$ = 2000 mJ m^{-2}

[g]$\gamma_{s\text{-}l}$ = 2400 mJ m^{-2}

[h]$\gamma_{s\text{-}l}$ = 2700 mJ m^{-2}

[i]$\gamma_{s\text{-}l}$ = 3000 mJ m^{-2}

[j]$\gamma_{s\text{-}l}$ = 1200 mJ m^{-2}

[k]$\gamma_{s\text{-}l}$ = 1490 mJ m^{-2}

[l]Sintered materials (porosity – 3–8%, content non-combined C – 0.50%)

[m]$\gamma_{s\text{-}l}$ = 1130 mJ m^{-2}

[n]Plasma-sprayed carbide coatings

[o]Metal impurities: Si-0.8%, Fe-0.3%, Ni-0.2%, Al-0.1%, Mg-0.1%, Ca-0.1%, Cu-0.1%, Ag-0.1%

[p]Metal impurities: Ag-0.2, Fe-0.1, Si-0.1, Mg-0.1%, Pb-0.1%

[q]Metal impurities: Si-0.3%, Mg-0.3%, Cu-0.2%, Ni-0.2%, Mn-0.2%, Ag-0.1%, C-0.003%, N-0.001%, O-0.001%

[r]Alloys prepared by refusion of carbonyl iron and graphite with spectral purity

[s]Fe–Cr alloy: Cr-65%, Fe-29%, C-6%

[t]Porosity ≤4%

[u]Fe–Si alloy: Si-75%, Fe-25%

[v]Porosity <4%, impurities <2 mas.%

[w]Porosity – 4–8%

[x]Porosity – 23%

[y]Metal impurities: Fe-1.0%, Al-0.3%, Mg-0.2%, Ag-0.2%, Ca-0.1%, Cu-0.1%

[z]Hot-pressed materials, poreless

Table 5.29 Diffusion rates and related parameters in the systems containing zirconium, carbon and zirconium carbides/carbonitrides phases at various temperatures[a,b]

Species pair	Temperature dependence of the diffusion coefficient (diffusivity) $D = D_0\exp[(-E_A/R)/T]$, $cm^2\ s^{-1}$	Temperature range, K (°C)	Remarks on materials characteristics and measurement method	Reference
C → α-Zr	$7.90 \times 10^{-4}\exp(-15{,}350/T)$	870–1070 (600–800)	Polycrystalline Zr, ^{14}C radiometric method	[26, 27, 1262]
	$(2.00 \pm 0.37) \times 10^{-3}\exp[(-18{,}200 \pm 300)/T]$	870–1120 (600–850)	Polycrystalline Zr (~2 mm grain size), ^{14}C radiometric (thin layer residual activity) method	[1, 1285, 1286]
	$3.51 \times 10^{-5}\exp(-15{,}500/T)$	870–1070 (600–800)	Polycrystalline Zr, ^{14}C radiometric method	[1, 32]
	$6.0 \times 10^{-5}\exp(-16{,}400/T)$	890–1010 (620–740)	Polycrystalline Zr, ^{14}C radiometric method	[1, 1288]
	$5.0 \times 10^{-8}\exp(-46{,}200/T)$	1010–1100 (740–830)	Polycrystalline Zr, ^{14}C radiometric method	[1, 1288]
	$6.02 \times 10^{-3}\exp(-23{,}400/T)$	700–1100 (430–830)	Calculated by using kinetic Monte-Carlo simulations	[1283]
C → β-Zr	$(8.90 \pm 1.60) \times 10^{-2}\exp[(-16{,}000 \pm 200)/T]$	1140–1520 (870–1250)	Polycrystalline Zr (~2 mm grain size), ^{14}C radiometric (thin layer residual activity) method	[1, 1285, 1286]
	$3.57 \times 10^{-2}\exp(-17{,}200/T)$	1370–1870 (1100–1600)	Polycrystalline 99.6%-purity Zr (contents: C – 0.01%, O – 0.11%, N – 0.035%, Fe – 0.22%, Hf – 0.015%), ^{14}C radiometric method	[1, 26, 27, 1259, 1260]
	$4.80 \times 10^{-3}\exp(-13{,}500/T)$	1170–1570 (900–1300)	Polycrystalline 99.9%-purity Zr (contents: N – 0.014%, Si – 0.05%, Fe – 0.04%, Hf – 0.07%), ^{14}C radiometric (diffusion couple) method[c]	[1, 26, 27, 1260, 1263, 1286, 1304]
	$1.30 \times 10^{-4}\exp(-12{,}700/T)$	1240–1820 (970–1550)	Polycrystalline Zr, metallography method	[1, 1289]

(continued)

Table 5.29 (continued)

Species pair	Temperature dependence of the diffusion coefficient (diffusivity) $D = D_0\exp[(-E_A/R)/T]$, $cm^2\ s^{-1}$	Temperature range, K (°C)	Remarks on materials characteristics and measurement method	Reference
	$3.44 \times 10^{-2}\exp[(-20{,}700 \pm 4{,}800)/T]$	1170–1570 (900–1300)	Parameters of reaction-chemical diffusion, contact saturation of solid Zr with C (graphite), metallography method (determination of the concentration gradient by microhardness)[c]	[1, 26, 27, 32, 42, 48, 1252, 1253, 1291]
	$4.52 \times 10^{3}\exp[(-9000 \pm 2{,}800)/T]$	1270–1870 (1000–1600)	Parameters of reaction-chemical diffusion, contact saturation of solid Zr with C (graphite), metallography method	[1250, 1251, 1254, 1294]
	$1.4 \times 10^{-6}\exp(-15{,}200/T)$	1370–1970 (1100–1700)	–	[1, 32, 1290]
	$10^{2}\exp(-44{,}500/T)$	1470–1670 (1200–1400)	Parameters of the diffusion saturation of solid Zr with C in vacuum	[1, 32, 1292]
	$1.6 \times 10^{-2}\exp(-34{,}600/T)$	1570–1770 (1300–1500)	Parameters of reaction-chemical diffusion, contact saturation of solid Zr with C (graphite)	[1, 1289]
	$2.4 \times 10^{-6}\exp(-20{,}100/T)$	1570–1820 (1300–1550)	Parameters of reaction-chemical diffusion, contact saturation of solid high-purity Zr (prepared by iodide method) with C	[1, 1289]
	$1.5 \times 10^{-6}\exp(-19{,}100/T)$	1570–1870 (1300–1600)	Parameters of reaction-chemical diffusion, contact saturation of solid Zr (prepared by arc-melting) with C	[1, 1289]
$C \rightarrow ZrC_{1-x}$	$1.32 \times 10^{2}\exp(-57{,}200/T)$ (for volume diffusion) $1.6\exp(-45{,}300/T)$ (for short-circuit diffusion)	1620–2420 (1350–2150)	Single crystal and polycrystalline ($0.2 \times 0.2 \times 3$ mm columnar grains) ZrC_{1-x} ($x = 0.03$, contents: non-combined C – 0.02%, Hf – 2.5%), ^{14}C radiometric method	[1, 4, 26, 27, 298, 1273, 1295, 1309, 1316–1318]
	$(3.3 \pm 0.6) \times 10^{2}\exp[(-57{,}400 \pm 3{,}000)/T]$	1870–2370 (1600–2100)	Polycrystalline (hot-pressed and annealed) ZrC_{1-x} ($x = 0.04$, porosity $\leq 5\%$, content non-combined C – 0.65%), ^{14}C radiometric method	[4, 26, 27, 298, 841, 1254, 1259, 1261, 1274]
	$56.4\exp[(-62{,}400 \pm 6{,}200)/T]$	2520–3010 (2250–2740)	Polycrystalline (hot-pressed and annealed) ZrC_{1-x} ($x = 0.19$, 110–130 μm grain size, porosity – 3%, contents: $O + N \leq 0.6\%$), ^{14}C radiometric method	[1, 4, 26, 27, 32, 298, 841, 1261]
	$(14.1^{+13.5}_{-12.7})\exp[(-54{,}800 \pm 3{,}000)/T]$	2670–3070 (2400–2800)	Polycrystalline (hot-pressed and annealed) ZrC_{1-x} ($x = 0.03$, $Zr(C_{0.97}N_{0.02}O_{0.01})_{0.97}$, 100–300 μm grain size, porosity 1–3%), ^{14}C radiometric (with sectioning technique) method	[1, 4, 26, 27, 298, 1272, 1316–1318]

(continued)

Table 5.29 (continued)

Species pair	Temperature dependence of the diffusion coefficient (diffusivity) $D = D_0\exp[(-E_A/R)/T]$, $cm^2\ s^{-1}$	Temperature range, K (°C)	Remarks on materials characteristics and measurement method	Reference
	4.26×10^{-9} 4.35×10^{-8} 2.88×10^{-7}	2500 (2230) 2800 (2530) 3100 (2830)	Calculated on the basis of developed statistic and thermodynamic theory of interstitial phases for ZrC_{1-x} ($x = 0.03$)	[337]
	$(19.5^{+40.5}_{-13.0})\exp[(-60{,}200 \pm 3{,}100)/T]$	2670–3070 (2400–2800)	Polycrystalline (hot-pressed and annealed) ZrC_{1-x} ($x = 0.07$, $Zr(C_{0.97}N_{0.02}O_{0.01})_{0.93}$, 100–300 μm grain size, porosity 1–3%), ^{14}C radiometric (with sectioning technique) method	[1, 4, 26, 27, 298, 1272]
	$(61.2^{+45.2}_{-54.0})\exp[(-64{,}200 \pm 6{,}000)/T]$	2670–3070 (2400–2800)	Polycrystalline (hot-pressed and annealed) ZrC_{1-x} ($x = 0.12$, $Zr(C_{0.88}N_{0.02}O_{0.11})_{0.88}$, 100–300 μm grain size, porosity 1–3%), ^{14}C radiometric (with sectioning technique) method	[1, 4, 26, 27, 298, 1272]
	$(2.64^{+4.26}_{-1.63}) \times 10^2\exp[(-67{,}000 \pm 4{,}200)/T]$	2670–3070 (2400–2800)	Polycrystalline (hot-pressed and annealed) ZrC_{1-x} ($x = 0.15$, $Zr(C_{0.95}N_{0.04}O_{0.01})_{0.85}$, 100–300 μm grain size, porosity 1–3%), ^{14}C radiometric (with sectioning technique) method	[1, 4, 26, 27, 298, 1272]
	$(3.16^{+8.80}_{-2.32}) \times 10^2\exp[(-69{,}200 \pm 3{,}800)/T]$	2670–3070 (2400–2800)	Polycrystalline (hot-pressed and annealed) ZrC_{1-x} ($x = 0.16$, $Zr(C_{0.96}N_{0.04}O_{0.01})_{0.84}$, 100–300 μm grain size, porosity 1–3%), ^{14}C radiometric (with sectioning technique) method	[1, 4, 26, 27, 298, 1272]
	$(2.8^{+5.0}_{-2.0}) \times 10^4\exp[(-82{,}600 \pm 2{,}600)/T]$	2770–3270 (2500–3000)	Polycrystalline (hot-pressed and annealed) ZrC_{1-x} ($x = 0.30$, $Zr(C_{0.98}N_{0.01}O_{0.01})_{0.70}$, porosity 1–3%), ^{14}C radiometric method	[4, 1295, 1298]
	$6.3 \times 10^3\exp(-N_C)\exp(-35{,}600/T)$	1570–2020 (1300–1750)	Chemical diffusion coefficient as a function of the atomic fraction of C (N_C) and temperature determined by diffusion couple technique through electron-probe microanalysis	[1319, 1320]
	$8.3 \times 10^{-3}\exp(9.2x)\exp[(-36{,}600 \pm 3{,}700)/T]$	1570–2020 (1300–1750)	Parameters of chemical diffusion ($0 < x \leq 0.49$) determined by diffusion couple technique through electron-probe microanalysis	[1295, 1296]
	$2.94 \times 10^2\exp(-54{,}700/T)$	1670–1870 (1400–1600)	^{14}C radiometric method	[1, 1289]
	$3.32\exp(-57{,}600/T)$	1870–2350 (1600–2080)	Parameters of chemical diffusion during solid-phase welding of pairs of Zr–ZrC_{1-x} materials	[1, 1293]

(continued)

Table 5.29 (continued)

Species pair	Temperature dependence of the diffusion coefficient (diffusivity) $D = D_0\exp[(-E_A/R)/T]$, $cm^2\ s^{-1}$	Temperature range, K (°C)	Remarks on materials characteristics and measurement method	Reference
	1.40×10^{-10} 5.09×10^{-9} 8.10×10^{-8}	2500 (2230) 2800 (2530) 3100 (2830)	Calculated on the basis of developed statistic and thermodynamic theory of interstitial phases for ZrC_{1-x} ($x = 0.30$)	[337]
	~exp(−46,000/*T*)	2070–2570 (1800–2300)	Rate of carburization of Zr obtained by the precipitation from gas phase on the surface of graphite	[26, 27, 1254, 1255]
	$5 \times 10^2\exp(-39{,}800/T)$	1920–2270 (1650–2000)	Rate of carburization of Zr obtained by the precipitation from gas phase on the surface of graphite; X-ray method	[26, 27, 1254, 1256, 1271]
	0.95exp(−39,600/*T*)	2270–3130 (2000–2860)	Parameters of temperature variation of ZrC_{1-x} layer growth rate constant upon C saturation of liquid Zr in graphite crucible, photomicrography of longitudinal sections[c]	[26, 27, 1254, 1257, 1258, 1304]
	1.83exp(−42,400/*T*)	2270–3070 (2000–2800)	Parameters of temperature variation of ZrC_{1-x} layer growth rate constant upon C saturation of liquid Zr in graphite crucible and ZrC_{1-x} recrystallization (solid-state grain growth); metallography method	[26, 27, 1266–1269]
	0.37exp(−38,400/*T*)	1570–2070 (1300–1800)	Parameters of reaction-chemical diffusion, contact saturation of solid Zr with C (graphite); metallography method	[26, 27, 1270]
	0.34exp(−34,700/*T*)	2170–2800 (1900–2530)	Chemical diffusion coefficients calculated on the basis of experiments on deposition of non-stoichiometric carbide ZrC_{1-x} ($0 < x \leq 0.44$, porosity $\leq 10\%$) from $CH_4 + H_2$ gas phase[d]	[1, 26, 27, 32, 1275]
	0.92exp(−40,500/*T*)	2510–2720 (2240–2450)	Chemical diffusion coefficients calculated on the basis of mass change measurements for the initial composition of ZrC_{1-x} with $x = 0.44$	[296]
	0.167exp(−41,700/*T*)	–	The values of chemical diffusion coefficients summarized on the basis of several works	[51, 296]
C → $(Zr_{1-y}Nb_y)C_{1-x}$	$(2.28^{+0.47}_{-0.30})\exp[(-48{,}600 \pm 1{,}300)/T]$	2770–3470 (2500–3200)	Polycrystalline $(Zr_{1-y}Nb_y)C_{1-x}$ ($x = 0.03$, $y = 0.52$), ^{14}C radiotracer method	[4, 337, 1295, 1298]
	8.26×10^{-9} 6.72×10^{-8} 3.56×10^{-7}	2500 (2230) 2800 (2530) 3100 (2830)	Calculated on the basis of developed statistic and thermodynamic theory of 3-component interstitial phases for $(Zr_{1-y}Nb_y)C_{1-x}$ ($x = 0.10$, $y = 0.52$)	[337]

(continued)

Table 5.29 (continued)

Species pair	Temperature dependence of the diffusion coefficient (diffusivity) $D = D_0\exp[(-E_A/R)/T]$, $cm^2\ s^{-1}$	Temperature range, K (°C)	Remarks on materials characteristics and measurement method	Reference
	$(0.84^{+0.47}_{-0.30})\exp[(-50{,}600 \pm 1{,}700)/T]$	2770–3470 (2500–3200)	Polycrystalline $(Zr_{1-y}Nb_y)C_{1-x}$ ($x = 0.18$, $y = 0.52$), ^{14}C radiotracer method	[4, 337, 1295, 1298]
	1.36×10^{-9} 1.20×10^{-8} 6.91×10^{-8}	2500 (2230) 2800 (2530) 3100 (2830)	Calculated on the basis of developed statistic and thermodynamic theory of 3-component interstitial phases for $(Zr_{1-y}Nb_y)C_{1-x}$ ($x = 0.20$, $y = 0.52$)	[337]
C → $(Zr_{1-y}U_y)C_{1-x}$	$9.0 \times 10^{-3}\exp(-32{,}200/T)$	2170–2570 (1900–2300)	Polycrystalline $(Zr_{1-y}U_y)C_{1-x}$ ($x = 0.06$, $y = 0.79$), ^{14}C radiotracer method	[1295, 1297]
	$2.4 \times 10^{-2}\exp(-36{,}200/T)$	2170–2570 (1900–2300)	Polycrystalline $(Zr_{1-y}U_y)C_{1-x}$ ($x = 0.07$, $y = 0.49$), ^{14}C radiotracer method	[1295, 1297]
	$10^{-2}\exp(-39{,}200/T)$	2170–2570 (1900–2300)	Polycrystalline $(Zr_{1-y}U_y)C_{1-x}$ ($x = 0.10$, $y = 0.28$), ^{14}C radiotracer method	[1295, 1297]
C → $ZrN_{1\pm x}$	$(1.59^{+0.27}_{-0.22}) \times 10^{-3}\exp(-46{,}400/T)$	2300–2960 (2030–2690)	Polycrystalline (nitrided through metal saturation) $ZrN_{0.95}$ (no porosity, content O $\leq$ 0.015%), ^{14}C radiometric method	[4, 26, 27, 1280]
Cs → ZrC_{1-x}	$2.3 \times 10^{-3}\exp(-38{,}200/T)$	1470–1870 (1200–1600)	Polycrystalline $ZrC_{\sim 1.0}$ (content non-combined C – 3.4%) coating (thickness – 45 μm) on 500 μm nuclear fuel particles (*see also* Table 5.27)	[4, 96, 781]
He → ZrC_{1-x}	$\sim\exp[(-8900 \pm 900)/T]$	1270–1870 (1000–1600)	Polycrystalline ZrC_{1-x} ($x = 0.08 \pm 0.02$, 1–5 μm grain size, porosity 1%), resonant $^3He(d,p)^4He$ nuclear reaction in as-implanted (3 MeV $^3He^+$) and thermally annealed materials	[1360]
	$5.13 \times 10^{-7}\exp(-9{,}050/T)$	200–2000 (−70–1730)	Parameters obtained from first-principles total energy and phonon frequency calculations (including the all atoms) on the basis of transition state theory (TST)	[808, 1361]
Nb → ZrC_{1-x}	$(1.81^{+3.30}_{-1.63}) \times 10^2\exp[(-84{,}600 \pm 4{,}300)/T]$	2770–3470 (2500–3200)	Polycrystalline (hot-pressed and annealed) ZrC_{1-x} ($x = 0.30$, $Zr(C_{0.98}N_{0.01}O_{0.01})_{0.70}$, porosity 1–3%), ^{95}Nb radiotracer method	[4, 1279, 1295, 1298]
Nb → $(Zr_{1-y}Nb_y)C_{1-x}$	$(0.51^{+1.33}_{-0.50}) \times 10^2\exp[(-77{,}100 \pm 4{,}900)/T]$	2770–3470 (2500–3200)	Polycrystalline $(Nb_{1-y}Zr_y)C_{1-x}$ ($0.03 \leq x \leq 0.18$, $y = 0.52$), ^{95}Nb radiotracer method[e]	[4, 337, 1298]

(continued)

Table 5.29 (continued)

Species pair	Temperature dependence of the diffusion coefficient (diffusivity) $D = D_0\exp[(-E_A/R)/T]$, $cm^2\ s^{-1}$	Temperature range, K (°C)	Remarks on materials characteristics and measurement method	Reference
	2.12×10^{-12} 5.71×10^{-11} 8.26×10^{-10}	2500 (2230) 2800 (2530) 3100 (2830)	Calculated on the basis of developed statistic and thermodynamic theory of 3-component interstitial phases for $(Nb_{1-y}Zr_y)C_{1-x}$ ($0.10 \leq x \leq 0.20$, $y = 0.52$)	[337]
Pm → ZrC_{1-x}	$(1.26^{+7.96}_{-1.10}) \times 10^2\exp[(-83{,}600 \pm 5{,}000)/T]$	2270–2820 (2000–2550)	Polycrystalline hot-pressed ZrC_{1-x} ($0.03 \leq x \leq 0.30$, contents: O + N ≈ 0.5%), ^{147}Pm radiotracer[e]	[4, 1282]
U → $(Zr_{1-y}U_y)C_{1-x}$	$4.6 \times 10^{-2}\exp(-52{,}800/T)$	2170–2570 (1900–2300)	Polycrystalline $(Zr_{1-y}U_y)C_{1-x}$ ($x = 0.06$, $y = 0.79$), ^{235}U radiotracer method	[1295, 1297]
	$1.6\exp(-62{,}900/T)$	2170–2570 (1900–2300)	Polycrystalline $(Zr_{1-y}U_y)C_{1-x}$ ($x = 0.07$, $y = 0.49$), ^{235}U radiotracer method	[1295, 1297]
	$9.6\exp(-68{,}000/T)$	2170–2570 (1900–2300)	Polycrystalline $(Zr_{1-y}U_y)C_{1-x}$ ($x = 0.10$, $y = 0.28$), ^{235}U radiotracer method	[1295, 1297]
W → ZrC_{1-x}	$(2.0^{+8.3}_{-1.6}) \times 10^4\exp[(-82{,}600 \pm 4{,}000)/T]$	2270–2820 (2000–2550)	Polycrystalline hot-pressed ZrC_{1-x} ($x \approx 0$, porosity – 5%, contents: O + N ≈ 0.5%), ^{185}W radiotracer method	[4, 1281]
Zr → ZrC_{1-x}	$(1.03^{+6.03}_{-0.88}) \times 10^3\exp[(-86{,}600 \pm 5{,}400)/T]$	2520–2920 (2250–2650)	Polycrystalline hot-pressed ZrC_{1-x} ($0.03 \leq x \leq 0.30$, porosity – 1–3%, contents: O + N ≤ 1.5–2.0%), ^{95}Zr radiotracer method[e]	[1, 4, 26, 27, 1272, 1279, 1295]
	$6.67 \times 10^5\exp(-81{,}700/T)$	–	Calculated on the basis of general theoretical approach on diffusion processes in transition metal monocarbides	[1, 32, 1278]
	9.93×10^{-13} 3.83×10^{-11} 7.62×10^{-10}	2500 (2230) 2800 (2530) 3100 (2830)	Calculated on the basis of developed statistic and thermodynamic theory of interstitial phases (for $0.03 \leq x \leq 0.30$)	[337]
	$2.4 \times 10^{10}\exp(-105{,}700/T)$	2670–3520 (2400–3250)	Effective diffusion coefficients calculated on the basis of sintering kinetics data of ZrC_{1-x} ($x = 0.18$) spherical particles	[26, 27, 1276, 1295]
	$10^{5.2-11.9x}\exp[-(96{,}400 - 137600x)/T]$	–	Effective diffusion coefficients calculated on the basis of data on several diffusion-controlled processes in sintered ZrC_{1-x} materials (steady-state creep in flexure for $0.10 \leq x \leq 0.37$, decarburization for $x = 0.38$ and carburization for $x = 0.44$)[f]	[26, 27, 1277]

(continued)

Table 5.29 (continued)

Species pair	Temperature dependence of the diffusion coefficient (diffusivity) $D = D_0\exp[(-E_A/R)/T]$, $cm^2\ s^{-1}$	Temperature range, K (°C)	Remarks on materials characteristics and measurement method	Reference
	~exp(−65,800/*T*)	–	Estimated on the basis of LeClaire's relation (connected with phase melting point)	[1254]
Zr → $(Zr_{1-y}U_y)C_{1-x}$	$7.1 \times 10^{-2}\exp(-53{,}900/T)$	2170–2570 (1900–2300)	Polycrystalline $(Zr_{1-y}U_y)C_{1-x}$ ($x = 0.06$, $y = 0.79$), ^{95}Zr radiotracer method	[1295, 1297]
	$0.23\exp(-57{,}400/T)$	2170–2570 (1900–2300)	Polycrystalline $(Zr_{1-y}U_y)C_{1-x}$ ($x = 0.07$, $y = 0.49$), ^{95}Zr radiotracer method	[1295, 1297]
	$1.3\exp(-61{,}900/T)$	2170–2570 (1900–2300)	Polycrystalline $(Zr_{1-y}U_y)C_{1-x}$ ($x = 0.10$, $y = 0.28$), ^{95}Zr radiotracer method	[1295, 1297]

[a]For the diffusion parameters of fission products *see* Table 5.27

[b]The chemical diffusion coefficient concentration and temperature dependencies can be represented by an equation of the form $D' = D'_0\exp(\beta x)\exp[(-E'_A/R)/T]$, where β is the concentration factor and x is the value of index in ZrC_{1-x} formula (for the ranges of $x = 0.02$–0.48 and $T = 1330$–2830 °C the recommended values are $D'_0 = 0.167\ cm^2s^{-1}$, $\beta = 9.2$ and $E'_A = 347\ kJ\ mol^{-1}$[296]), and the discrepancy between chemical and self-diffusivities may be estimated on the basis of the thermodynamical relationship: $D' = D \times [1 + (d\ln\gamma_C/d\ln N_C)]$, where γ_C is the activity coefficient of carbon and N_C is the atomic fraction of carbon (evaluated at 2000 °C in ZrC_{1-x}: at $x = 0.01$, $d\ln\gamma_C/d\ln N_C = 107$ and at $x = 0.20$, $d\ln\gamma_C/d\ln N_C = 20$ [1299]); the approximate values of apparent activation energy for some diffusion controlled processes in ZrC_{1-x}: (**a**) collective recrystallization of loosely-poured powders – 180 kJ mol^{-1}($ZrC_{0.65–0.73}$, 1600–2000 °C) [3, 1305], 210 kJ mol^{-1}($ZrC_{0.80}$, 1600–2000 °C) [3, 1305], 220 kJ mol^{-1}($ZrC_{0.97}$, 1600–2000 °C) [3, 1305], 250 ± 2 kJ mol^{-1}($ZrC_{\sim1.0}$, 1800–2000 °C) [4, 1300, 1301], 260 ± 4 kJ mol^{-1}($ZrC_{\sim1.0}$, 1400–2100 °C) [1303]; (**b**) recrystallization (grain growth) – 180 kJ mol^{-1}(sintered after surface cold-working $ZrC_{0.60}$, 1600–2000 °C) [3, 1306, 1307, 1312, 1315], 190 kJ mol^{-1}(sintered after surface cold-working $ZrC_{0.73}$, 1600–2000 °C) [3, 1306, 1307, 1312, 1315], 205 kJ mol^{-1}(sintered after surface cold-working $ZrC_{0.80}$, 1600–2000 °C) [3, 1306, 1307, 1312, 1315], 80 kJ mol^{-1}(sintered $ZrC_{0.85\pm0.05}$, 2200–2700 °C) [1349], 290 ± 14 kJ mol^{-1}(sintered $ZrC_{0.92}$, 2400–2600 °C) [4, 1300, 1301], 235 kJ mol^{-1}(sintered after surface cold-working $ZrC_{0.97}$, 1600–2000 °C) [3, 1306, 1307, 1312, 1315], 400 (initial stage) and 75 (final stage) kJ mol^{-1}(sintered and 25%-deformed at 2000 °C $ZrC_{0.98}$, 2500–2600 °C) [742], 140 kJ mol^{-1}(sintered after bulk deformation $ZrC_{\sim1.0}$, ≤ 1900 °C) [4, 1302]; (**c**) powder sintering densification – 365–395 kJ mol^{-1}($ZrC_{0.68}$, 1450–2000 °C) [62, 1313], 360–380 kJ mol^{-1}($ZrC_{0.79}$, 1450–2000 °C) [62, 1313], 355–370 kJ mol^{-1}($ZrC_{0.91}$, 1450–2000 °C) [62, 1313], 345–360 kJ mol^{-1}($ZrC_{0.99}$, 1450–2000 °C) [62, 1313], (**d**) powder hot-pressing densification – 125 kJ mol^{-1}($ZrC_{0.65}$, mean particle size 10–18 μm, 2000–2300 °C) [1308], 135 kJ mol^{-1}($ZrC_{0.70}$, mean particle size 10–18 μm, 2000–2300 °C) [1308], 145 kJ mol^{-1}($ZrC_{0.75}$, mean particle size 10–18 μm, 2000–2300 °C) [1308], 155 kJ mol^{-1}($ZrC_{0.80}$, mean particle size 10–18 μm, 2000–2300 °C) [1308], 180 kJ

(continued)

Table 5.29 (continued)

mol^{-1}($ZrC_{0.90}$, mean particle size 10–18 μm, 2000–2300 °C) [1308], 170 kJ mol^{-1}($ZrC_{0.96}$, mean particle size 3–5 μm, specific surface area 0.5 m^2g^{-1}, 1700–2400 °C) [1321], 190 kJ mol^{-1}($ZrC_{\sim 1.0}$, mean particle size 10–18 μm, 2000–2300 °C) [1308]; (**e**) powder spark-plasma sintering densification – 690–770 kJ mol^{-1}($ZrC_{0.94}O_{0.05}$, mean particle size ~0.5 μm, specific surface area 1.15 m^2g^{-1}, 1900–2200 °C) [630, 1075]; (**f**) internal friction peaks characteristics (Marx-Wert formula) – 390 kJ mol^{-1}(sintered $ZrC_{0.94}$, 1730 °C) [1329], data on creep – *see* Sect. 5.4 (Table 5.16), *see also* Sect. 5.5 (Table 5.21)

[c]The diffusion coefficients were assumed to be independent on concentration

[d]The parameters probably contain significant contribution of surface and grain-boundary diffusion due to the rather high porosity of the materials

[e]Temperature dependences of the diffusion coefficients are the same for all the values of x in the indicated interval

[f]The given equation shows that in these processes the diffusion flows of both metallic and carbon components of ZrC_{1-x} phases contribute: the metallic component contribution is more considerable for near-stoichiometric compositions, the carbon component contribution – for the phases with higher values of x formula index; the difference in the effective diffusion coefficients for various ZrC_{1-x} phases decreases with an increase of temperature, it agrees with the assumption [1278] of approximate equality of the self-diffusion coefficients for the both components near carbide-graphite eutectic point

Table 5.30 The interaction of near-stoichiometric zirconium monocarbide ZrC_{1-x} materials with some common chemical reagents in aqueous solutions or molten states [1, 10, 11, 30–32, 35, 39, 42, 47, 96, 213, 223, 452, 607, 781, 979, 1124, 1158, 1330, 1331, 1333–1337, 1346, 1367, 1368]

Reagent, formula (density or concentration of aqueous solution)[a]	Treatment conditions		Character of interaction[b]
	Temperature, °C	Exposure time, h	
HCl (1:1)	20	24	Decomposes up to 2%
	100	1	Decomposes up to 1%
	110	2	Decomposes up to 5%
HCl (d = 1.19)	20	24	Decomposes up to 2%
	100–110	1	Decomposes up to 2–4%
6M HCl	100	40	Very slight interaction
0.05M H_2SO_4	20	1–600	Corrosion rate (coatings on steel) – 2.1×10^{-5} mg cm^{-2} s
H_2SO_4 (1:4)	20	24	Decomposes up to 2%
	100	1	Decomposes up to 5%
	140	1	Decomposes up to 24%
H_2SO_4 (d = 1.84)	20	24	Decomposes up to 3%
	100	1	Decomposes up to 9%
	200–235	1.5	Decomposes completely with the formation of oxysulfates and amorphous C; composition of released gases: CH_4 71, CO_2 26.5, H_2 2.5 mol.% and SO_2 (deposition of amorphous C – 7%)[c]
	260	1	Decomposes completely with the formation of oxysulfates and amorphous C; composition of released gases: CH_4 48, CO_2 48, H_2 3 mol.% and SO_2 (deposition of amorphous C – 14%)[c]
HNO_3 (1:1)	20	24	Decomposes up to 24%
	100	1	Decomposes up to 18%
	110	1–2	Decomposes with the formation of soluble nitrates up to 94%; composition of released gases: CO_2 95–97, H_2 2–3 mol.% and NO/NO_2 (no precipitated C, solution is transparent)[d]
HNO_3 (d = 1.37)	100	1	Decomposes up to 56%
HNO_3 (d = 1.43)	20	24	Decomposes up to 17%
	110–120	0.5	Decomposes completely with the formation of soluble nitrates; composition of released gases: CO_2 95–97, H_2 2–3 mol.%, CH_4 – traces and NO/NO_2 (no precipitated C, solution is transparent)[d]

(continued)

Table 5.30 (continued)

Reagent, formula (density or concentration of aqueous solution)[a]	Treatment conditions		Character of interaction[b]
	Temperature, °C	Exposure time, h	
HF (d = 1.15)	20	24	Decomposes completely
10M HF	100	~50	Decomposes, composition of released gases: CH_4 92, H_2 6, C_2H_6 0.6 and CO_2 0.35 mol.%[e,f]
H_3PO_4 (1:3)	20	24	Decomposes up to 4%
	≥100	2	Decomposes up to 12%
H_3PO_4 (d = 1.70)	20	24	Decomposes up to 2%
	250–255	1	Decomposes completely with the formation (or deposition) of $ZrO(H_2PO_4)_2$; composition of released gases: CH_4 91.4, CO_2 7.8 and H_2 0.8 mol.%[g]
	260–270	1	Decomposes completely with the formation (or deposition) of $ZrO(H_2PO_4)_2$; composition of released gases: CH_4 87.3, CO_2 12 and H_2 0.7 mol.%[g]
$HClO_4$ (1:3)	20	24	Decomposes up to 1%
	≥100	2	Decomposes up to 16%
$HClO_4$ (d = 1.35)	20	24	Decomposes up to 3%
	≥100	2	Decomposes up to 98%
$H_2C_2O_4$[h] (saturated solution)	20	24	Decomposes up to 2%
	100–104	2	Decomposes up to 8%
Other organic acids	110	1	No decomposition
NaOH (10%)	20	24	Decomposes up to 9%
	≥100	2	No decomposition
NaOH (20%)	20	24	No decomposition
	100	1	No decomposition
	110	2	No decomposition
4 M NaOH	100	20	No reaction
NaCl (3%)	20	1–600	Corrosion rate (coatings on steel) – 2.6×10^{-6} mg cm^{-2} s^{-1}
Na_2CO_3 (10%)	20	1–600	Corrosion rate (coatings on steel) – 2.4×10^{-7} mg cm^{-2} s^{-1}
KOH (30%)	20	1–600	Corrosion rate (coatings on steel) – 4.0×10^{-7} mg cm^{-2} s^{-1}
H_2O (technical)	20	1–600	Corrosion rate (coatings on steel) – 1.0×10^{-6} mg cm^{-2} s^{-1}
NaOH (fused)	650	–	Decomposes
Na_2CO_3 (fused)	850	–	Decomposes
KOH (fused)	500	5	Decomposes, composition of released gases: H_2 86, CH_4 9, O_2 0.5, CO_2 0.10, CO 0.07 and C_2H_2 0.02 mol.% (He gas carrier)[e,i]

(continued)

Table 5.30 (continued)

Reagent, formula (density or concentration of aqueous solution)[a]	Treatment conditions		Character of interaction[b]
	Temperature, °C	Exposure time, h	
$KClO_3$ (fused)	360	–	Decomposes easily
KNO_3 (fused)	340	–	Decomposes easily
K_2MnO_4 (fused)	190	–	Decomposes easily
NaCl + $BaCl_2$ (fused)	1100	–	Weak interaction
H_2O_2 (30%)	100	1	Decomposes up to 1%
	110	2	Decomposes up to 3%
NH_4F (5%)	110	2	Decomposes up to 24%
NH_4F (10%)	100	1	Decomposes up to 7%
$(NH_4)_2S_2O_8$ (25%)	100–110	1	Decomposes up to ~60–80% (C deficient phases decomposes completely)
HCl (1:1) + H_2SO_4 (1:4)	110	1	Decomposes partially
3HCl (1:1) + HNO_3 (1:2)	110	1	Decomposes up to ~90%
3HCl (d = 1.19) + HNO_3 (d = 1.43)	20	24	Decomposes up to 86%
	100–105	1–2	Decomposes up to 94% (C deficient phases decomposes completely)
HCl (d = 1.19) + H_2O_2 (30%)	110	1	Decomposes up to ~50–60%
HCl (d = 1.19) + $(NH_4)_2S_2O_8$ (25%)	110	1	Decomposes up to ~50–60%
HCl (d = 1.19) + Br_2 (HBrO, HBr) (saturated solution)	–	–	Decomposes up to ~90%
H_2SO_4 (d = 1.84) + HNO_3 (d = 1.43)	20	24	Decomposes up to 98%
	140	2	Decomposes completely
H_2SO_4 (1:4) + H_3PO_4 (1:3)	20	24	Decomposes partially with hydrolysis and salt depositions
	≥150	2	Decomposes partially with hydrolysis and salt depositions
H_2SO_4 (d = 1.84) + H_3PO_4 (1:1)	20	2	Decomposes up to 3%
	≥150	2	Decomposes partially
H_2SO_4 (d = 1.84) + H_3PO_4 (d = 1.70)	20	24	Decomposes up to 3%
	240	2	Mainly decomposes with the formation of salt depositions
H_2SO_4 (1:4) + $H_2C_2O_4$[h] (saturated solution)	20	24	Decomposes up to 4%
	180	2	Decomposes up to 9%
H_2SO_4 (d = 1.84) + $H_2C_2O_4$[h] (saturated solution)	20	24	Decomposes up to 4%
	≥100	2	Decomposes completely
H_2SO_4 (1:4) + H_2O_2 (30%)	110	1	Decomposes up to 99%[j]
H_2SO_4 (1:4) + $(NH_4)_2S_2O_8$ (25%)	100–110	1	Decomposes up to 96% (C deficient phases decomposes completely)
H_2SO_4 (1:1) + $(NH_4)_2S_2O_8$ (25%)	110	1	Decomposes up to 98%
6M H_2SO_4 + 0.5M CrO_3	100	~50	Decomposes, composition of released gases: CO_2 94, CO 4, O_2 1, H_2 0.08 and CH_4 0.04 mol.%[e,k]

(continued)

Table 5.30 (continued)

Reagent, formula (density or concentration of aqueous solution)[a]	Treatment conditions		Character of interaction[b]
	Temperature, °C	Exposure time, h	
1M H_2SO_4 + 1M H_2O_2	20	~200	Decomposes, composition of released gases: O_2 95, CO 3, CO_2 1, CH_4 0.4, H_2 0.07 and C_2H_2 0.03 mol.%[e]
0.5M H_2SO_4 + 1M $CuSO_4$	20	200	No reaction
HNO_3 (d = 1.43) + HF (d = 1.15)[l]	20	24	Decomposes completely
	110	0.5	Decomposes completely
4HNO_3 (d = 1.43) + HF (d = 1.15)	20	24	Decomposes completely
	110	0.5	Decomposes completely
HNO_3 (d = 1.43) + NH_4F (5%)	110–120	0.5	Decomposes
H_2O_2 (30%) + $H_2C_2O_4^h$ (50%)	110	1	Hydrolysis
H_2O_2 (30%) + $H_2C_4H_4O_6^m$ (50%)	110	1	Decomposes up to 4%
H_2O_2 (30%) + $H_3C_6H_5O_7^n$ (50%)	110	1	Decomposes up to 5%
H_2O_2 (30%) + $H_4C_{10}H_{12}N_2O_8^o$ (saturated solution)	110	1	Decomposes up to 2%
H_2O_2 (30%) + NH_4F	–	–	Decomposes completely
$(NH_4)_2S_2O_8$ + $H_4C_{10}H_{12}N_2O_8^o$	–	–	Decomposes up to 2%
$(NH_4)_2S_2O_8$ + NH_4F	–	–	Decomposes up to 99%
4NaOH (20%) + Br_2 (HBrO, HBr)	20	24	Decomposes up to ~2–7%
	105	2	Decomposes up to 13%
4NaOH (20%) + H_2O_2 (30%)	20	24	Decomposes up to 47%
	112	2	Decomposes up to 97%
4NaOH (20%) + $K_3[Fe(CN)_6]$ (10%)	20	24	Decomposes up to 18%
	110	2	Decomposes up to 63%
CCl_4 + Br_2[p]	20	200	No reaction

[a]All the ratios are given in volume fractions and percents in mass (weight)
[b]When it is not indicated specially, the character reported is related to the powders with mean particle size of 40–50 μm
[c]The decomposition of ZrC_{1-x} starts at 200–220 °C, at 230–250 °C the dissolution is very energetical and decomposition, accompanied by the release of amorphous finely-dispersed C, mainly finishes in 30–40 min; it is represented by the main reaction equations:
$ZrC_{1-x} + 2H_2SO_4 = Zr(SO_4)_2 + (1 - x)CH_4 + 2xH_2$,
$ZrC_{1-x} + 4H_2SO_4 = Zr(SO_4)_2 + (1 - x)C + 2SO_2 + 4H_2O$
as well as secondary reaction equations:
$C + 2H_2SO_4 = CO_2 + 2H_2O + 2SO_2$,
$CH_4 + 4H_2SO_4 = CO_2 + 6H_2O + 4SO_2$
[d]Represented by the following reaction equation:
$ZrC_{1-x} + 6HNO_3 = ZrO(NO_3)_2 + (1 - x)CO_2 + 2(1 - x)NO + 2(1 + x)NO_2 + 3H_2O$
[e]Sintered cylinders (diameter 12.7 mm, height 6.4 mm)
[f]The interaction with hot HF solution may be described by the equation
$ZrC_{1-x} + 6HF = ZrF_6^{-2} + (1 - x)CH_4 + 2(1 + 2x)H^+$
[g]Represented by the following reaction equation:
$ZrC_{1-x} + 2H_3PO_4 + H_2O = ZrO(H_2PO_4)_2 + (1 - x)CH_4 + 2xH_2$
[h]Oxalic acid
[i]The decomposition reaction may be represented by the following schemes:
$ZrC_{1-x} + KOH \rightarrow ZrO_{2-\delta}{\cdot}yK_2O + K_2CO_3 + 4H_2\uparrow + K_2O$,
$ZrC_{1-x} + KOH \rightarrow ZrO_{2-\delta}{\cdot}yK_2O + CH_4\uparrow + K_2O$

(continued)

Table 5.30 (continued)

[j]The resistance of ZrC_{1-x} phases in similar solvents increases with increasing their deviation from the stoichiometry
[k]The decomposition reaction may be represented by the following chemical equations:
$3ZrC_{1-x} + 8CrO_3 + 30H^+ = 3ZrO^{+2} + 3(1 - x)CO_2 + 8Cr^{+3} + 15H_2O$,
$ZrC_{1-x} + 2CrO_3 + 8H^+ = ZrO^{+2} + (1 - x)CO + 2Cr^{+3} + 4H_2O$
[l]Recommended chemical etching (dissolving) agents for ZrC_{1-x}: (**a**) 25–50 ml HNO_3 (d = 1.43) + 0.25–1.0 ml HF (d = 1.15) + 50 ml H_2O at 60–80 °C (for the SEM observation of materials with excess carbon) [781, 1337]; (**b**) 7 ml HF (d = 1.15) + 3 ml HNO_3 (d = 1.43) [785]; (**c**) 3 ml HNO_3 (d = 1.43) + 2.0 ml HF (d = 1.15) + 95 ml H_2O (for the SEM observation of materials) [1338]; (**d**) 1 ml HF (d = 1.15) + 1 ml HNO_3 (d = 1.43) + 2 ml H_2O with 1 min exposure at room temperature (for cleavage planes) [1339, 1340]; (**e**) the same with 1–10 s exposure (for electroetching with voltage – 1–10 V) [1339]; (**f**) 25 ml HNO_3 (d = 1.43) + 25 ml $HClO_4$ (d = 1.35) (for the SEM observation of materials with excess carbon) [1337]; (**g**) 2 ml HNO_3 (d = 1.43) + 1 ml HCl (d = 1.19) + 1 ml H_2O with HF (<5%) traces at a constant temperature <35 °C (for the electrochemical treatment of carbide emitters, cathode – Ta wire, DC voltage in pulsed operation – 15–30 V) [525]; (**h**) 2 ml HNO_3 (d = 1.43) + 0.5 ml NaF (0.1 N solution) + 3 ml H_2O at a constant temperature <35 °C (for the electrochemical treatment of carbide emitters, cathode – Ta wire, DC voltage in continuous or pulsed operation – 30 V) [525]; (**i**) 3 ml HNO_3 (d = 1.43) + 1 ml HCl (d = 1.19) + ~2 ml H_2O at a constant temperature <35 °C (for the electrochemical treatment of carbide emitters, cathode – Ta wire, DC voltage in continuous operation initially and pulsed operation just before drop-off – 25–35 V) [525]; (**j**) 10 g $K_3[Fe(CN)_6]$ + 2–10 g KOH/NaOH + 100 ml H_2O (Murakami's reagent for metallographic analysis) [879]; (**k**) 5% Br_2 solution in CH_3OH (methanol) (for the separation of carbide phases from metal-carbide alloys) [879]; (**l**) boiling H_2SO_4 (d = 1.84) [171]; (**m**) HNO_3 (d = 1.43) + 2–3% Na_2SiF_6 with 5 min exposure at room or cooler temperatures [1341]
[m]Tartaric acid
[n]Citric acid
[o]Ethylenediaminetetraacetic acid (EDTA)
[p]Organic solutions

Table 5.31 The interaction of zirconium oxycarbide $ZrC_{1-x}O_y$ phases (powdered materials with mean particle size of 15–20 μm, treated at 100 °C, 1 h exposure) with some common chemical reagents in aqueous solutions [979]

Reagent, formula (density or concentration of aqueous solution)[a]	Decomposition, %		
	$ZrC_{0.82}O^b_{0.12}$	$ZrC_{0.77}O^b_{0.16}$	$ZrC_{0.70}O^b_{0.24}$
HCl (1:1)	0	2	2
HCl (d = 1.19)	2	2	3
H_2SO_4 (1:4)	8	21	67
H_2SO_4 (d = 1.84)	13	24	66
HNO_3 (1:1)	10	6	3
HNO_3 (d = 1.37)	17	12	4
3HCl (1:1) + HNO_3 (1:1)	100	44	16
$(NH_4)_2S_2O_8$ (25%)	97	70	3
H_2SO_4 (1:4) + $(NH_4)_2S_2O_8$ (25%)	100	100	98
NH_4F (10%)	23	38	86
H_2O_2 (30%)	1	4	3
NaOH (20%)	1	0	0

[a]All the ratios are given in volume fractions and percents in mass (weight)
[b]Non-combined C is not detected

In comparison with other ultra-high temperature materials the summarized data on the chemical behaviour of zirconium monocarbide are given in Addendum.

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Addendum

A.1 Structures

Summarized general data on the structural properties (atomic or molecular weights, phase homogeneity regions, crystal systems, types, space groups and lattice parameters, calculated and experimental densities) of the high-melting elements (carbon and refractory metals) and refractory compounds (carbides of Hf, Nb, Ta and Zr), which were considered separately and comprehensively in the main chapters of the first and second volumes of the book, are given (in alphabetical order) in Table A.1 [1–52, 67, 68, 90].

A.2 Thermal Properties

The most important thermodynamic properties (standard heat of formation, standard molar entropy, molar and specific heat capacities, molar and specific enthalpies (heats) of melting (fusion) and vaporization, molar and specific enthalpy differences $H_T - H_{298}$) of the ultra-high temperature materials considered separately and comprehensively in the first and second volumes of the book are summarized in Tables A.2 and A.3 [1–3, 7–9, 13–20, 28–33, 36–41, 43, 44, 53–66, 71, 85, 90]. For the general comparison, some other thermal properties (melting and boiling points, coefficients of linear thermal expansion, relative thermal expansion, thermal conductivity, vapour pressure and vaporization rate) of ultra-high temperature materials (graphite, refractory metals and Hf, Nb, Ta, Zr carbides) are given in Tables A.4 and

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A.5 [1–3, 6–21, 23–32, 35–44, 46, 47, 50–57, 66, 69, 70, 72–75, 79–90] . The values of the heat capacities, enthalpy differences, thermal expansion and thermal conductivity properties and vaporizations parameters of the materials are presented there in the wide range from room (or moderate) to ultra-high temperatures.

A.3 Electro-magnetic and Optical Properties

For the general comparison, the main electro-magnetic and optical properties (specific electrical resistance, temperature coefficient of electroresistance, integral and monochromatic emittances, thermionic emission characteristics and molar magnetic susceptibility) of carbon (graphite), refractory metals and Hf, Nb, Ta, Zr carbides are given in Table A.6 [1, 3, 4, 6–9, 11–15, 18–21, 23, 31–41, 43–47, 56, 76–78, 90–92]. The values of the electrical resistance and integral and monochromatic emittances of the high-melting elements and compounds are summarized in the wide range from room (or moderate) to ultra-high temperatures.

A.4 Physico-mechanical Properties

The relative comparison of some physico-mechanical properties (hardness, tensile and compressive strengths and Young's modulus) of ultra-high temperature materials (graphite, refractory metals and Hf, Nb, Ta, Zr carbides), which were considered above separately and comprehensively in the main chapters of the first and second volumes of the book, can be carried out on the basis of the data presented in Table A.7 [1, 4, 6–15, 18–21, 23, 29–41, 43–47, 57, 83, 90, 93, 94]. The values of the hardness, strength and elasticity of the ultra-high temperature materials are summarized there for the wide range from room to ultra-high temperatures.

A.5 Nuclear Physical Properties

Nuclear physical properties (isotopic mass range, total number of isotopes, thermal neutron macroscopic and microscopiccross sections, moderating ability and capture resonance integral) of the elements (carbon and refractory metals) and Hf, Nb, Ta, Zr carbides are summarized in Table A.8 [1, 2, 95–100].

A.6 Chemical Properties

For the ultra-high temperature materials (elements and compounds) considered in details in the first and second volumes of the book, the characters of interaction with some chemical reagents in the aqueous solutions at room temperature are summarized in Table A.9 [2, 3, 7, 9, 11, 13, 17, 19, 23, 32, 38, 43, 44, 56, 90, 101–104, 220–229]. For the cubic carbides of Hf, Nb, Ta, Zr (NaCl-structured), considered in the second volume, the interaction with isomorphic refractory compounds (carbides, nitrides, oxides, phosphides and sulphides) is considered in Table A.10 [1, 8, 14–21, 24, 28–33, 35–44, 48–50, 64, 67, 90, 230–235].

A.7 Porosity-Property Relationships

The methods of powder metallurgy and particulate technologies, such as conventional sintering, hot pressing, hot isostatic pressing, spark plasma sintering and others, are widely spread for the preparation and production of ultra-high temperature materials, especially for those based on Hf, Nb, Ta, Zr refractory carbides considered in this volume of the book. The effect of porosity on physical properties (thermal conductivity, electrical resistance, strength and elastic characteristics) of sintered metallic and ceramic materials has been the subject of numerous experimental and theoretical studies. Some of the formalae proposed by various researchers for the descrption of physical properties and characteristics of the materials with residual porosity are given in Table A.11 [105–219]. The diversity of the formulae is associated with the necessity to take into account the shape, size and number of pores, the surface conditions of the test samples, the grain size in the materials, the degree of their local inhomogeneity, the presence of residual stresses and some other factors.

Table A.1 Structural properties (atomic or molecular weight, phase homogeneity region, system, type and space group of crystal structure, calculated and experimental densities)

Materials formula	Atomic or molecular weight	Phase homogeneity region[a]	Crystal structure						Density, g cm^{-3}	
			System	Type	Space group	Lattice parameters[b], nm			Z[c]	Calculated (XRD)	Experimental[d]
						a	*c*	*c/a*			
					Elements						
C (graphite)[e]	12.01	–	Hexagonal	Graphite	$P6_3/mmc$	0.2464	0.6711	2.7236	4	2.267	2.26
Ir	192.22	–	Cubic (fcc)	Cu	$Fm(-3)m$	0.3839	–	–	4	22.56	22.45
Mo	95.94	–	Cubic (bcc)	W	$Im(-3)m$	0.3147	–	–	2	10.22	10.24
Nb	92.91	–	Cubic (bcc)	W	$Im(-3)m$	0.3301	–	–	2	8.578	8.59
Os	190.23	–	Hexagonal (hcp)	Mg	$P6_3/mmc$	0.2734	0.4320	1.580	2	22.60	22.48
Re	186.21	–	Hexagonal (hcp)	Mg	$P6_3/mmc$	0.2762	0.4457	1.614	2	21.01	21.00
Ta	180.95	–	Cubic (bcc)	W	$Im(-3)m$	0.3303	–	–	2	16.68	16.60
W	183.84	–	Cubic (bcc)	W	$Im(-3)m$	0.3165	–	–	2	19.25	19.20
					Carbides						
HfC_{1-x}	184.74–190.26	$0.02 \le x \le 0.48$	Cubic (fcc)	NaCl	$Fm(-3)m$	0.4640	–	–	4	12.65	12.60
$Nb_{2+x}C$	200.62–262.87	$0.03 \le x \le 0.70$	Hexagonal[f]	NiAs	$P6_3/mmc$	0.3127	0.4972	1.590	1	7.80	7.80
NbC_{1-x}	100.60–104.44	$0.04 \le x \le 0.36$	Cubic (fcc)	NaCl	$Fm(-3)m$	0.4470	–	–	4	7.80	7.75
$Ta_{2\pm x}C$	341.34–527.72	$-0.18 \le x \le 0.85$	Hexagonal[g]	NiAs	$P6_3/mmc$	0.3105	0.4935	1.589	1	15.05	14.90
TaC_{1-x}	187.80–192.72	$0.02 \le x \le 0.43$	Cubic (fcc)	NaCl	$Fm(-3)m$	0.4457	–	–	4	14.50	14.45
ZrC_{1-x}	97.95–102.99	$0.02 \le x \le 0.44$	Cubic (fcc)	NaCl	$Fm(-3)m$	0.4700	–	–	4	6.60	6.60

[a]The max. variation in the non-metal content of refractory compounds within their temperature intervals of thermal stability according to the phase diagrams
[b]Concerning the refractory compounds, it is given for the compositions with the min. deviation from the stoichiometry
[c]Number of formula units in the unit cell
[d]Pycnometric density
[e]2H-Graphite (or α-carbon, α-graphite)
[f]The data are given for the high-temperature modification (Nb semicarbide phases undergo the phase transformations: α-$Nb_2C \leftrightarrow$ β-$Nb_{2+x}C$ at ~1200–1230 °C and β-$Nb_{2+x}C \leftrightarrow$ γ-$Nb_{2\pm x}C$ at ~2450–2575 °C, the temperatures of transformations depend on the phase compositions, *see* Sect. 4.1)
[g]The data are given for the high-temperature modification (Ta semicarbide phases undergo the phase transformation: α-$Ta_{2+x}C \leftrightarrow$ β-$Ta_{2\pm x}C$ at ~1930–2270 °C, the temperatures of transformations depend on the phase compositions, *see* Sect. 2.1)

Table A.2 Thermodynamic properties (standard heat of formation, standard molar entropy, molar and specific heat capacities)

Materials formula	Standard heat of formation $-\Delta H°_{298}$, J mol^{-1} K^{-1}	Standard molar entropy $S°_{298}$, J mol^{-1} K^{-1}	Molar heat capacity c_p, J mol^{-1} K^{-1}, at temperatures, K (°C)					Specific heat capacity c, J kg^{-1} K^{-1}, at temperatures, K (°C)				
			298 (25)	1000 (730)	2000 (1730)	3000 (2730)	4000 (3730)	298 (25)	1000 (730)	2000 (1730)	3000 (2730)	4000 (3730)
						Elements						
C (graphite)	–	5.74	8.5	21.0	25.0	26.5	–	700	1750	2100	2200	–
Ir	–	35.5	25.0	29.2	35.2	–	–	130	153	191	–	–
Mo	–	28.6	24.0	28.7	36.9	41.8	–	250	300	385	435	–
Nb	–	36.5	24.7	27.7	31.8	33.5	–	265	300	340	360	–
Os	–	32.6	24.9	27.5	31.2	34.9	37.7	130	145	164	183	198
Re	–	36.5	25.5	29.0	34.9	42.3	46.0	138	156	188	227	247
Ta	–	41.5	25.4	27.9	31.0	44.1	–	140	154	171	244	–
W	–	32.8	24.3	27.6	32.3	37.2	41.8	132	150	176	202	227
						Carbides[a]						
HfC_{1-x}	235	41.2	37.5	51.6	58.0	63.6	69.2	197	271	305	336	364
$Nb_{2+x}C$	195	64.0	63.5	81.1	88.3	94.2	–	317	404	440	470	–
NbC_{1-x}	141	37.3	36.8	51.2	60.0	66.7	72.5	352	490	575	639	694
$Ta_{2\pm x}C$	208	81.6	61.0	79.5	94.1	108.1	–	163	213	252	289	–
TaC_{1-x}	145	42.3	36.8	51.1	59.1	66.9	74.2	191	267	307	347	385
ZrC_{1-x}	197	33.3	37.9	53.4	57.4	60.9	76.2	368	518	557	591	740

[a]All the data are given for the pure near-stoichiometric compositions

Table A.3 Thermodynamic properties (molar and specific enthalpies (heats) of melting/vaporization, molar and specific enthalpy differences)

Materials formula	Enthalpy (heat) of melting[a]		Enthalpy (heat) of vaporization[b]		Molar enthalpy differences $H_T - H_{298}$, kJ mol^{-1}, at temperatures, K (°C)						Specific enthalpy differences $H_T - H_{298}$, kJ kg^{-1}, at temperatures, K (°C)					
	Molar, kJ mol^{-1}	Specific, MJ kg^{-1}	Molar, MJ mol^{-1}	Specific, MJ kg^{-1}	1000 (730)	2000 (1730)	3000 (2730)	4000 (3730)	5000 (4730)	6000 (5730)	1000 (730)	2000 (1730)	3000 (2730)	4000 (3730)	5000 (4730)	6000 (5730)
Elements																
C (graphite)	146	12.2	0.71	59.1	12.8	36.6	62.8	–	–	–	1067	3050	5230	–	–	–
Ir	41.1	0.214	0.60	3.14	18.9	52.0	118.2	160.1	–	–	98.3	271	615	833	–	–
Mo	37.5	0.290	0.60	6.16	18.5	51.0	123.6	165.4	799	856	192.2	531	1288	1723	8324	8920
Nb	30.0	0.315	0.69	7.32	18.5	48.2	108.2	141.7	175	892	198.5	518	1164	1525	1884	9598
Os	57.9	0.304	0.73	3.84	18.4	47.7	80.7	149.4	187	962	96.6	251	424	786	983	5058
Re	34.1	0.178	0.70	3.74	19.2	51.0	89.5	167.9	214	–	103.0	274	481	902	1149	–
Ta	36.6	0.200	0.75	4.15	19.0	48.2	83.6	150.6	186	–	104.9	266	462	832	1029	–
W	52.3	0.220	0.80	4.33	18.3	48.1	83.8	158.6	200	–	99.3	262	456	867	1090	–
Carbides[c]																
HfC_{1-x}	–	–	1.57[d]	8.24[d]	32.7	87.7	148.5	214.8[e]	–	–	171.8	461	780	1128	–	–
$Nb_{2+x}C$	–	–	–	–	52.4	139.7	229.6[f]	–	–	–	261.2	696	1144	–	–	–
NbC_{1-x}	92.0[g]	0.881[g]	~0.7[h]	~6.7[h]	32.6	88.4	151.4	313.6	386	458	312.3	847	1450	2998	3697	4387
$Ta_{2\pm x}C$	–	–	2.48[d]	6.63[d]	51.0	137.9	239.0	353.7[e]	–	–	136.4	369	639	946	–	–
TaC_{1-x}	105.0[g]	0.545[g]	–	–	32.5	87.7	150.6	221.0	285	352	168.7	455	782	1147	1479	1827
ZrC_{1-x}	83.7[g,i]	0.813[g]	0.61[d,j]	5.92[d]	34.3	89.9	149.0	291.0	408	487	333.0	873	1447	2825	3961	4728

(continued)

Table A.3 (continued)

[a]For refractory metals molar and specific enthalpies (heats) of melting are given at the melting points

[b]For refractory metals molar and specific enthalpies (heats) of vaporization are given at the boiling points

[c]All the data are given for the pure near-stoichiometric compositions

[d]At 298.15 K

[e]Extrapolated data

[f]Extrapolated data (for middle-temperature modification β-$Nb_{2+x}C$)

[g]At the melting point

[h]The approximate value for the vaporization (dissociation) process of NbC_{1-x} (crystal) = NbC_{1-x-y} (crystal) + yC (gas) at 2800 K (2530 °C)

[i]Molar enthalpy (heat) of melting at 298.15 K is 79.5 kJ mol^{-1}

[j]Molar enthalpy (heat) of vaporization (dissociation) is 1.52 kJ mol^{-1} (the value was measured on the basis of Langmuir mode, but also calculated in accordance to the second law of thermodynamics as a sum of average partial enthalpies (heats) of vaporization of C_1 (gas) and Zr (gas) from $ZrC_{\sim 1.0}$)

Table A.4 Thermal properties (melting and boiling points, coefficients of linear thermal expansion, relative thermal expansion and thermal conductivity)

Materials formula	Melting point, K (°C)	Boiling point, K (°C)	Average coefficient of linear thermal expansion, 10^{-6} K^{-1}		Relative thermal expansion, %, affected by heating from room temperature to *T*, K (°C)					Thermal conductivity, W m^{-1} K^{-1}, at temperature, K (°C)					
			At room temperature	20–1700 °C	1000 (730)	1500 (1230)	2000 (1730)	2500 (2230)	3000 (2730)	290 (20)	1000 (730)	1500 (1230)	2000 (1730)	2500 (2230)	3000 (2730)
							Elements								
C (graphite[a])	4200 (3930)[b]	4200 (3930)[b]	2.5–3.5	5.1–5.8	0.25–0.30	0.50–0.58	0.80–0.92	1.13–1.28	1.47–1.70	100–200	60–70	45–60	35–50	35–45	35–45
C (graphite[c])	4200 (3930)[b]	4200 (3930)[b]	0–20	1.5–25	0.08–1.30	0.19–2.40	0.32–3.60	0.49–4.90	–	3–1000	1.5–200	1–150	0.7–100	0.5–70	–
Ir	2720 (2450)	4700 (4430)	6.8	9.2	0.52	0.98	1.55	2.18[d]	–	145	130	115[e]	110	100[d]	–
Mo	2890 (2620)	5100 (4830)	5.4	6.7	0.38	0.74	1.15	1.66	–	135	115	95	80	70	–
Nb	2740 (2470)	5020 (4750)	7.1	8.9	0.56	1.02	1.52	2.00	–	50	60	70	80	90	–
Os	3320 (3050)	5300 (5030)	6.1	~9	0.32 (?)	–	–	–	–	88[f]	85[f]	84[f]	85[f]	~80	–
Re	3450 (3180)	5870 (5600)	6.6	7.0	0.47	0.85	–	–	–	70	56[g]	50	45	42	–
Ta	3270 (3000)	5700 (5430)	6.5	7.7	0.50	0.88	1.32	1.82	2.40	60	70	76	80	84	–
W	3680 (3410)	6000 (5730)	4.3	5.0	0.32	0.52	0.88	1.20	1.60	160	120	105	100	95	92
							Carbides[h]								
HfC_{1-x}	4220 (3950)	5670 (5400)	5.7	7.0	0.47	0.82	1.19	1.56	1.98	15	21	27	33	39	45
$Nb_{2+x}C$	3350 (3080)	–	3.4/9.9/5.7[i]	8.6/9.8[j]	–	–	–	–	–	12 (?)	–	–	–	–	–
NbC_{1-x}	3870 (3600)	5275 (5000)	6.4	7.3	0.46	0.84	1.28	1.77	2.32	18	25	32	38	43	47
$Ta_{2\pm x}C$	3610 (3335)	5740 (5465)	4.2/2.7[k]	–	–	–	–	–	–	35	–	–	–	–	–
TaC_{1-x}	4260 (3990)	5770 (5500)	5.8	7.0	0.44	0.80	1.20	1.65	2.14	25	30	34	37	39	41
ZrC_{1-x}	3760 (3490)	5370 (5100)	4.5	7.1	0.45	0.85	1.30	1.77	2.27	35	25	30	32	35	40

(continued)

Table A.4 (continued)

[a]High quality pure industrial polycrystalline (quasi-isotropic) graphite

[b]Total gas pressure over the solid/liquid surface – 10 MPa

[c]Highly oriented pyrolitic graphite (HOPG)

[d]Extrapolated values

[e]Interpolated value

[f]Average values; in different directions the minimal and maximal values are 76 and 111 W m^{-1} K^{-1} at 1000 (730) K (°C), 75 and 110 W m^{-1} K^{-1} at 1500 (1230) K (°C), 77 and 107 W m^{-1} K^{-1} at 2000 (1730) K (°C)

[g]Average value; in different directions the minimal and maximal values are 54 and 60 W m^{-1} K^{-1}

[h]All the data are given for the pure near-stoichiometric compositions

[i]The experimental data for the low-temperature modification α-Nb_2C are given for its main crystallographic directions: *a*/*b*/*c*, respectively

[j]The experimental data for the middle-temperature modification β-$Nb_{2+x}C$ given for its main crystallographic directions (for *a* – in the numerator and for *c* – in the denominator, respectively) were extrapolated to the region of higher temperatures

[k]The experimental data for the low-temperature modification α-$Ta_{2+x}C$ are given for its main crystallographic directions: for *a* – in the numerator and for *c* – in the denominator, respectively

Table A.5 Thermal properties (vapour pressure and vaporization rate)

Materials formula	Temperature, K (°C), corresponding to vapour pressure over materials surface, Pa							Temperature, K (°C), corresponding to mass vaporization rate, kg m^{-2} s^{-1}					Temperature, K (°C), corresponding to linear vaporization rate, m s^{-1}				
	10^{-7}	10^{-5}	10^{-3}	0.1	10	10^{3}	10^{5}	10^{-10}	10^{-8}	10^{-6}	10^{-4}	10^{-2}	10^{-14}	10^{-12}	10^{-10}	10^{-8}	10^{-6}
								Elements									
C (graphite)	1790 (1520)	1950 (1680)	2160 (1890)	2410 (2140)	2740 (2470)	3170 (2900)	3790 (3520)	2000 (1730)	2200 (1930)	2450 (2180)	2800 (2530)	3300 (3030)	1950 (1680)	2130 (1860)	2360 (2090)	2700 (2430)	3100 (2830)
Ir	1750 (1480)	1940 (1670)	2190 (1920)	2510 (2240)	–	–	–	–	–	–	–	–	–	–	–	–	–
Mo	1750 (1480)	1960 (1690)	2210 (1940)	2550 (2280)	3010 (2740)	3770 (3500)	~5100 (~4830)	1750 (1480)	1960 (1690)	2230 (1960)	2560 (2290)	3020 (2750)	1750 (1480)	1960 (1690)	2220 (1950)	2560 (2290)	3030 (2760)
Nb	1930 (1660)	2120 (1850)	2380 (2110)	2720 (2450)	3220 (2950)	3940 (3670)	~4530 (~4260)	1930 (1660)	2130 (1860)	2400 (2130)	2750 (2480)	3250 (2980)	1930 (1660)	2130 (1860)	2390 (2120)	2740 (2470)	3230 (2960)
Os	~1880 (~1610)	~2090 (~1820)	~2370 (~2100)	~2700 (~2430)	~3160 (~2890)	~3850 (~3580)	~4880 (~4610)	~1870 (~1600)	~2080 (~1810)	~2350 (~2080)	~2700 (~2430)	>3150 (>2880)	~1910 (~1640)	~2120 (~1850)	~2410 (~2140)	~2780 (~2510)	~3270 (~3000)
Re	2070 (1800)	2320 (2050)	2630 (2360)	3040 (2770)	3610 (3340)	4460 (4190)	~5920 (~5650)	2060 (1790)	2300 (2030)	2620 (2350)	3040 (2770)	–	2100 (1830)	2350 (2080)	2680 (2410)	3120 (2850)	–
Ta	2110 (1840)	2350 (2080)	2650 (2380)	3050 (2780)	3590 (3320)	4350 (4080)	~5700 (~5430)	2100 (1830)	2340 (2070)	2650 (2380)	3020 (2750)	–	2130 (1860)	2360 (2090)	2670 (2400)	3070 (2800)	–
W	2250 (1980)	2500 (2230)	2820 (2550)	3220 (2950)	3760 (3490)	4520 (4250)	5800 (5530)	2230 (1960)	2480 (2210)	2810 (2540)	3220 (2950)	–	2270 (2000)	2520 (2250)	2860 (2590)	3290 (3020)	–

(continued)

Table A.5 (continued)

Materials formula	Temperature, K (°C), corresponding to vapour pressure over materials surface, Pa							Temperature, K (°C), corresponding to mass vaporization rate, kg m^{-2} s^{-1}					Temperature, K (°C), corresponding to linear vaporization rate, m s^{-1}				
	10^{-7}	10^{-5}	10^{-3}	0.1	10	10^{3}	10^{5}	10^{-10}	10^{-8}	10^{-6}	10^{-4}	10^{-2}	10^{-14}	10^{-12}	10^{-10}	10^{-8}	10^{-6}
								Carbides[a]									
HfC_{1-x}	2080 (1810)	2270 (2000)	2550 (2280)	2920 (2650)	3410 (3140)	–	~5670 (~5400)	2070 (1800)	2260 (1990)	2530 (2260)	2910 (2640)	3400 (3130)	2030 (1760)	2270 (2000)	2550 (2280)	2920 (2650)	3410 (3140)
$Nb_{2+x}C$[b]	–	–	–	–	–	–	–	–	–	–	–	–	–	–	–	–	–
NbC_{1-x}	~1790 (~1520)	~2050 (~1780)	~2400 (~2130)	2770 (2500)	3570 (3300)	–	~5275 (~5000)	~1710 (~1440)	2020 (1750)	2460 (2190)	~3140 (~2870)	–	~1690 (~1420)	1990 (1720)	2440 (2170)	~3120 (~2850)	–
$Ta_{2\pm x}C$	–	–	–	–	–	–	–	–	~2420 (~2150)	~2730 (~2460)	~3130 (~2860)	~3660 (~3390)	–	~2440 (~2170)	~2760 (~2490)	~3170 (~2900)	~3710 (~3440)
TaC_{1-x} (C/Ta)[c]	~1890 (~1620)/ /2490 (2220)[c]	~2140 (~1870)/ /2760 (2490)[c]	2470 (2200)/ /3100 (2830)[c]	2910 (2640)/ /3550 (3280)[c]	~3540 (~3270)	–	–	~1940 (~1670)	2210 (1940)	2570 (2300)	~3070 (~2800)	–	~1950 (~1680)	2220 (1950)	2590 (2320)	~3090 (~2820)	–
ZrC_{1-x} (C/Zr)[d]	~1970 (~1700)/ /~1950 (~1680)[d]	~2190 (~1920)/ /~2170 (~1900)[d]	2480 (2210)/ /2460 (2190)[d]	2850 (2580)/ /2810 (2540)[d]	~3340 (~3070)/ /~3280 (~3010)[d]	–	~5370 (~5100)	~1960 (~1690)	2190 (1920)	2460 (2190)	~2820 (~2550)	~3330 (~3060)	~1940 (~1670)	2160 (1890)	2440 (2170)	~2790 (~2520)	~3250 (~2980)

[a]All the data are given for the pure near-stoichiometric compositions

[b]No data on the vaporization of $Nb_{2+x}C$ phases are available

[c]The vapour pressures over TaC_{1-x} materials surface are given separately: in the numerator for C and in the denominator for Ta

[d]The vapour pressures over ZrC_{1-x} materials surface are given separately: in the numerator for C and in the denominator for Zr

Table A.6 Electro-magnetic and optical properties (specific electrical resistance, thermal coefficient of electroresistance, integral and monochromatic emittances, thermionic emission characteristics and molar magnetic susceptibility)

Materials formula	Specific electrical resistance (resistivity), μΩ m, at temperature, K (°C)						Thermal coef. of electro-resistance at 20-1700 °C, 10^{-3} K^{-1}	Integral emittance ε_T[a] at temperature, K (°C)				Monochromatic emittance (spectral emissivity) ε_λ (λ = 0.665 μm)[a] at temp., K (°C)				Thermionic emission characteristics		Molar magnetic susceptibility χ_{mol}(SI)[b], 10^{-6} cm^3 mol^{-1}
	290 (20)	1000 (730)	1500 (1230)	2000 (1730)	2500 (2230)	3000 (2730)		1000 (730)	2000 (1730)	2500 (2230)	3000 (2730)	1000 (730)	2000 (1730)	2500 (2230)	3000 (2730)	Electron work function, eV	Richardson constant, 10^4 A m^{-2} K^{-1}	
Elements																		
C (graphite[c])	10	8.5	9.5	11	12	13	–	0.59–0.78	0.72–0.84	–	–	0.78–0.94	0.75–0.89	0.74–0.88	–	4.0–5.0	~15–60	–
C (graphite[d])	0.5–10^4	2.5–3.5×10^3	2.5–2×10^3	3.5–1.3×10^3	4.0–0.7×10^3	–	–	0.60–0.75	0.73–0.86	–	–	0.77–0.92	0.77–0.92	–	–	–	–	(–3200) –(–75)
Ir	0.05	0.20	0.34	0.48	> 0.6	–	5.1	–	~0.18[e]	0.21	–	–	~0.25[e]	0.215	–	4.7–5.7	100–120	+315
Mo	0.055	0.22	0.36	0.45	0.55	–	3.6	0.06	0.21	0.26	–	0.39	0.35	0.34	–	4.0–5.0	55–115	+905
Nb	0.15	0.43	0.55	0.70	0.83	–	1.9	~0.10	0.22	0.25	–	0.38	0.37	0.36	–	3.9–4.9	~35–70	+2615
Os	0.09	0.38[f]	0.55	–	–	–	~4.2	–	–	–	–	–	~0.4	~0.4	–	4.7–5.9	~10^5 (?)	+140
Re	0.18	0.64	0.84	1.0	1.1	–	2.4	~0.15	~0.25	0.29	0.33	0.43	0.40	0.38	0.36	4.7–5.3	~50–700	+840
Ta	0.135	0.44	0.62	0.79	0.95	1.1	2.8	~0.10	0.23	0.28	0.33	0.46	0.42	0.40	0.38	3.9–4.8	~40–120	+1935
W	0.055	0.245	0.40	0.57	0.75	0.925	5.5	0.11	0.26	0.30	0.33	0.45	0.43	0.42	0.41	4.2–5.8	60–200	+665
Carbides[g]																		
HfC_{1-x}	0.45	0.85	1.2	1.5	1.8	2.2	1.4	~0.25[h]	0.44	0.45	0.48	~0.55	0.65	0.70	~0.75	2.0–4.5	15–230	–400
$Nb_{2+x}C$	1.4 (?)	–	–	–	–	–	–	–	–	–	–	–	–	–	–	4.1–5.2	–	+500
NbC_{1-x}	0.45	0.75	1.0	1.2	1.4	1.7	0.85	0.42	0.44	0.46	0.50	~0.66[h]	0.63	0.62	0.61	2.2–4.2	70–400	+200
$Ta_{2\pm x}C$	0.45	–	–	–	–	–	–	–	–	–	–	–	–	–	–	4.1–4.7	–	>+200
TaC_{1-x}	0.40	0.75	0.95	1.2	1.5	1.75	1.1	~0.30	0.36	0.37	0.44	0.53	0.48	0.42	0.39	3.1–4.7	0.2–140	+200
ZrC_{1-x}	0.40	1.0	1.5	1.8	2.2	2.6	1.2	~0.68	0.59	0.55	~0.50	~0.60[h]	0.48	~0.42[i]	~0.40[i]	2.1–4.8	1–360	–200

(continued)

Table A.6 (contined)

[a]Measured on non-oxidized surfaces
[b]Measured at room temperature
[c]High quality industrial polycrystalline (quasi-isotropic) graphite with high purity
[d]Highly oriented pyrolitic graphite (HOPG)
[e]Interpolated data
[f]Average value; in different directions the minimal and maximal values are 270 and 430 nΩ m, respectively
[g]All the data are given for the pure near-stoichiometric compositions
[h]The experimental data were extrapolated to the region of lower temperatures
[i]The experimental data were extrapolated to the region of higher temperatures

Table A.7 Physico-mechanical properties (hardness, tensile and compressive strengths and Young's modulus)

Materials formula	Hardness[a] *HV*, GPa, at temperature, K (°C)					Ultimate tensile strength[b], MPa, at temperature, K (°C)						Ultimate compressive strength, MPa, at temperature, K (°C)					Young's modulus, GPa, at temperature, K (°C)				
	290 (20)	1000 (730)	1500 (1230)	2000 (1730)	2500 (2230)	290 (20)	1000 (730)	1500 (1230)	2000 (1730)	2500 (2230)	3000 (2730)	290 (20)	1000 (730)	1500 (1230)	2000 (1730)	2500 (2230)	290 (20)	1000 (730)	1500 (1230)	2000 (1730)	2500 (2230)
										Elements											
C (graphite[c])	0.3–1.0	0.3–1.0	0.3–1.0	0.4–1.0	0.4–1.0	6.4–10.0	7.8–15.0	8.8–18.0	10–21	~22	–	68	70	70	75	95	2.6–8.0	3.3–8.5	4.0–9.5	5.1–10.2	6– 11.5
C (graphite[d])	1.4–3.7	–	–	–	–	5–170	~140[e]	~160[e]	120–170[e]	150–190[e]	~350[e]	120–470	150–470	170–470	~200	–	~3–40	10–28[e]	~26[e]	23–24[e]	18–21[e]
Ir	2.0–17.2	~1	–	–	–	500–1100	–	–	~75	~20	–	–	–	–	–	–	525–540	–	<420	–	–
Mo	1.6–4.0	~0.7	~0.45	~0.14	~0.045	250–800	120–500	60–200	25–80	~7	–	–	–	–	–	–	320–330	280–300	~250	~200	–
Nb	0.7–2.5	~0.7	~0.12	~0.04	~0.006	200–700	160–300	~60	~20	–	–	–	–	–	–	–	104–110	~105	~105	–	–
Os	3.0–10.0[f]	~1.9–4.0[f]	~1.3–4.0[f]	–	–	2000–7200[f]	–	–	–	–	–	1000–3600	–	–	–	–	~560	~500	–	–	–
Re	1.3–8.0	~2.0[g]	~1.25	–	–	500–2300	500–1200	400–700	300–400	50–100	~20	–	–	–	–	–	460–520	~390	~330	–	–
Ta	0.4–3.0	~0.5[g]	~0.2	–	–	200–1000	~190	~85	~35	~15	<6	–	–	–	–	–	~185	–	–	–	–
W	1.4–6.0	0.9–1.6	0.6–1.2	0.3–0.5	~0.1	300–1900	180–800	100–350	60–140	35–45	~15	–	–	–	–	–	400–415	370	350	330	250

(continued)

Table A.7 (continued)

Materials formula	Hardness[a] *HV*, GPa, at temperature, K (°C)					Ultimate tensile strength[b], MPa, at temperature, K (°C)						Ultimate compressive strength, MPa, at temperature, K (°C)					Young's modulus, GPa, at temperature, K (°C)				
	290 (20)	1000 (730)	1500 (1230)	2000 (1730)	2500 (2230)	290 (20)	1000 (730)	1500 (1230)	2000 (1730)	2500 (2230)	3000 (2730)	290 (20)	1000 (730)	1500 (1230)	2000 (1730)	2500 (2230)	290 (20)	1000 (730)	1500 (1230)	2000 (1730)	2500 (2230)
											Carbides[h]										
HfC_{1-x}	16.0–29.0	5.0–9.0	2.0–4.0	2.0	~1.5[i]	240	180	130	100	70	30	–	–	–	–	–	320–480	270–320	230–300	180–280	100–250
$Nb_{2+x}C$	20.5–25.8	–	–	–	–	–	–	–	–	–	–	–	–	–	–	–	~410[j]	–	–	–	–
NbC_{1-x}	15.0–24.5	5.0–9.0	1.0–3.5	0.3–2.0	~0.1	75–240	70–160	60–100	~60	~50	10–35	1500–2500	~1000	400	~200	~150	280–580	250–500	230–470	210–460	180–280[i]
$Ta_{2\pm x}C$	9.5–18.0	–	–	–	–	230–275[k]	–	–	–	–	–	–	–	–	–	–	440–530	~400	~380	~320	–
TaC_{1-x}	13.0–25.0	7.0–15.0	2.5–6.0	1.0–3.5	0.5–1.0[i]	80–240	80–180	80–150	80–120	80–110	40–60	800–1800	–	–	–	–	290–560	390–500	350–470	330–440	300–400
ZrC_{1-x}	16.0–29.0	4.0–9.0	0.5–4.0	0.2–1.5	~1.0	100–300	70–210	70–250	40–270	20–250	10–160	800–2900	500–1200	300–1000	~600	~150	350–500	350–480	340–450	320–420	270–350

[a]For refractory metals the minimal values of hardness range are given for annealed specimens and maximal values – for hardened (cold-worked) specimens
[b]For refractory metals the minimal values of ultimate tensile strength range are given for annealed specimens and maximal – for hardened (cold-worked) specimens
[c]Polycrystalline (quasi-isotropic) graphite
[d]Highly oriented pyrolitic graphite (HOPG)
[e]In parallel direction to deposition plane
[f]Hardness and ultimate tensile strength of Os varies in magnitude also depending on crystal orientation
[g]For hardened (cold-worked) specimens
[h]All the data are given for the pure near-stoichiometric compositions
[i]The experimental data were extrapolated to the region of higher temperatures
[j]The DFT-calculated values: for the low-temperature modification α-Nb_2C – 220–430 GPa and for the middle-temperature modification β-$Nb_{2+x}C$ – 310 GPa
[k]Estimated on the basis of flexural (bending) strength measurements

Table A.8 Nuclear physical properties (isotopic mass range, total number of isotopes, thermal neutron macroscopic cross sections, moderating ability and capture resonance integral)

Materials formula	Isotopic mass range (minimal and maximal mass numbers)	Total number of isotopes (stable or nearly stable / radioactive / metastable isomers)	Macroscopic (microscopic) thermal neutron cross sections[a], cm^{-1} (b)		Neutron moderating ability (macroscopic slowing-down power)[b] $\xi\Sigma_s$, 10^{-3}	Resonance integral for neutron capture I_γ, b
			capture (absorption) Σ_a (σ_a)	scattering[c] Σ_s (σ_s)		
			Elements			
C	8–22	15 (2/13/0)	~0.32 × 10^{-3} (3.50 × 10^{-3})	~0.43 (~4.75)	~70	1.57 × 10^{-3}
Ir	164–199	68 (2/34/32)	29.9 (425.3)	0.984 (14.0)	–	2150
Mo[d]	83–115	41 (7/26/7)	0.163 (2.51–2.55)	0.368 (5.71)	9.27	24
Nb (^{93}Nb)	81–113	58 (1/32/25)	0.064 (1.15)	0.348 (6.255)	5.82	8.5
Os	162–197	45 (7/28/10)	1.14 (16.0)	1.05 (14.7)	8.2	180
Re[e]	160–194	56 (2/33/21)	6.09 (89.7)	0.781 (11.5)	14.2	831
Ta (^{181}Ta)	155–190	73 (1/35/37)	1.14 (20.6)	0.333 (6.01)	3.04	660
W[f]	158–192	46 (5/30/11)	1.16 (18.4)	0.292 (4.60)	3.35	352
			Carbides[g]			
HfC_{1-x}	153–188 / 8–22	63 (5/31/27)[h]	4.17 (52.6)[i]	~0.59 (7.5)[i]	–	–
$Nb_{2+x}C$	81–113 / 8–22	58 (1/32/25)[h]	0.055 (0.77)[j]	~0.41 (5.8)[j]	–	–
NbC_{1-x}	81–113 / 8–22	58 (1/32/25)[h]	0.052 (0.59)[i]	~0.49 (5.5)[i]	–	–
$Ta_{2\pm x}C$	155–190 / 8–22	73 (1/35/37)[h]	1.00 (13.7)[j]	~0.41 (5.6)[j]	–	–
TaC_{1-x}	155–190 / 8–22	73 (1/35/37)[h]	0.93 (10.4)[i]	~0.48 (5.4)[i]	–	–
ZrC_{1-x}	78–110 / 8–22	39 (5/28/6)[h]	7.27 × 10^{-3} (0.095)[i]	~0.43 (5.6)[i]	35.8	–

[a]For 2200 m s^{-1} neutrons

[b]ξ – the average logarithmic decrement (or the average change in the logarithm of neutron energy per collision)

Table A.8 (contined)

[c]Total bound scattering cross sections
[d]For ^{98}Mo: σ_a = 0.127 b and σ_s = 5.44 b, for ^{96}Mo: σ_a = 0.5 b and σ_s = 4.83 b, for ^{95}Mo: σ_a = 13.1 b and σ_s = 6.5 b, for ^{92}Mo: σ_a = 0.019 b and σ_s = 6 b
[e]For ^{187}Re σ_a = 76.4 b and σ_s = 11.9 b, for ^{185}Re σ_a = 112 b and σ_s = 10.7 b
[f]For ^{184}W: σ_a = 1.7 b and σ_s = 7.03 b, for ^{186}W: σ_a = 37.9 b and σ_s = 0.065 b, for ^{182}W: σ_a = 20.7 b and σ_s = 6.1 b, for ^{183}W: σ_a = 10.1 b and σ_s = 5.7 b, for ^{180}W: σ_a = 30 b and σ_s = 3 b
[g]All the data are given for the pure near-stoichiometric compositions
[h]The data are given for the metals forming refractory compounds
[i]The formula $\sigma_i = [\sigma_{Me} + (1 - x)\sigma_C]/(2 - x)$ was used to calculate the normalized thermal neutron cross sections for monocarbide MeC_{1-x} compositions
[j]The formula $\sigma_i = [(2 \pm x)\sigma_{Me} + \sigma_C]/(3 \pm x)$ was used to calculate the normalized thermal neutron cross sections for semicarbide $Me_{2\pm x}C$ compositions

Table A.9 Chemical properties (character of interaction with some chemical reagents in the aqueous solutions at room temperature)

Materials formula	Chemical reagents (formulae, concentrations)[a]													
	HCl		H_2SO_4		HNO_3		H_3PO_4	HF	HCl + HNO_3	HF + HNO_3	NH_4OH	NaOH	KOH	H_2O_2
	Diluted	Conc.	Diluted	Conc.	Diluted	Conc.								
							Elements							
C (graphite)	No	No	No	No	Slight[b]	Oxid.	Slight[b]	No	Oxid.	Oxid.	No	No	No	Slight[b]
Ir	No	No	No	No	No	No	No	No	No	No	No	No[c]	No[c]	No
Mo	No	Slight[d]	No	Slight[d]	Slight	Dissol.[e]	No[f]	Dissol.	Slight[e]	Dissol.[g]	No[h]	Slight[i]	No[j]	Oxid.
Nb	No	No	No	No	No	Slight	No	Dissol.	No	Dissol.	No	No[k]	No[k]	No
Os	No	No	No[l]	No[l]	No[l]	Slight[m]	–	No	Slight[m]	–	No	No[c]	No	–
Re	No	No	Slight	Dissol.	Dissol.	Dissol.	–	No	Dissol.	Dissol.	Slight	No[n]	No[n]	Oxid.[n]
Ta	No	No	No	No	No	No	No	Slight	Slight	Dissol.[g]	No	No[o]	No[o]	No
W	No	No[p]	No	No[p]	Slight	Oxid.	No[p]	No[q]	Slight[m]	Dissol.	No[h]	No[j]	No[j]	Oxid.
							Carbides[r]							
HfC_{1-x}	No[s]	No	No[s]	No[t]	Slight[t]	Slight[t]	Slight[t]	Decomp.	Decomp.	Decomp.	–	No	–	No
$Nb_{2+x}C$	No	No	No	No	–	–	–	Decomp.	–	–	–	–	–	No
NbC_{1-x}	No	No[s]	No	No[t]	Slight	No[s]	Slight	Decomp.	Slight[t]	Decomp.	–	No	No	Decomp.
$Ta_{2\pm x}C$	No	No	No	No	–	–	–	Decomp.	–	–	–	–	–	No
TaC_{1-x}	No	No	No	No[t]	No	No	No	Decomp.	No	Decomp.	No	No	No	No[s]
ZrC_{1-x}	No	No	No[s]	No[t]	Slight[t]	Slight[t]	Slight[t]	Decomp.	Decomp.	Decomp.	–	No[u]	No[u]	No

[a]The accepted abbreviations: Conc.—concentrated, Oxid.—oxidation, Dissol.—dissolution, Decomp.—decomposition
[b]Oxidized electrochemically
[c]The slight interaction in the melts

Table A.9 (contined)

[d]The slight dissolution while heating
[e]By heating the dissolution intensifies considerably
[f]The interaction in the presence of O_2 or some oxidizers
[g]The intensive interaction is observed
[h]The interaction initiates only in the presence of H_2O_2
[i]The intensive interaction in the presence of oxidizers
[j]The interaction with the melts in the presence of O_2
[k]The oxidation in the melts
[l]The interaction in the presence of O_2
[m]The dissolution while heating
[n]The interaction while heating
[o]The interaction in the melts
[p]The slight interaction while heating
[q]The dissolution in the presence of oxidizers
[r]All the data are given for the pure near-stoichiometric compositions
[s]The partial decomposition while heating
[t]The complete decomposition while heating
[u]The partial decomposition in the melts

Table A.10 Chemical properties (the interaction character of isomorphic refractory compounds with cubic NaCl structure)

Materials formula	Refractory compounds (with cubic NaCl structure)[a]													
						Carbides								
	HfC_{1-x}	α-MoC_{1-x}	NbC_{1-x}	NpC_{1-x}	PuC_{1-x}	ScC_{1-x}	TaC_{1-x}	$ThC_{1\pm x}$	TiC_{1-x}	$UC_{1\pm x}$	VC_{1-x}	γ-WC_{1-x}	YC_{1-x}	ZrC_{1-x}
HfC_{1-x}	–	Cont.[b]	Cont.	cont.	cont.	cont.	Cont.	Term.	Cont.	Cont.	Term.	Cont.[c]	term.	Cont.
NbC_{1-x}	Cont.	Cont.[d]	–	term.	Cont.	cont.	Cont.	Term.	Cont.	Cont.	Cont.	Cont.[c]	term.	Cont.
TaC_{1-x}	Cont.	Cont.[d]	Cont.	term.	cont.	cont.	–	Term.	Cont.	Cont.	Cont.	Cont.[c]	term.	Cont.
ZrC_{1-x}	Cont.	Cont.[d]	Cont.	cont.	cont.	cont.	Cont.	Term.	Cont.	Cont.	Term.	Cont.[e]	cont.	–
						Nitrides								
	$CeN_{1\pm x}$	$DyN_{1\pm x}$	$HfN_{1\pm x}$	NbN_{1-x}	NpN_{1-x}	PuN_{1-x}	ScN_{1-x}	TaN_{1-x}	$ThN_{1\pm x}$	$TiN_{1\pm x}$	UN_{1-x}	VN_{1-x}	$YN_{1\pm x}$	$ZrN_{1\pm x}$
HfC_{1-x}	term.	cont.	Cont.	Cont.	cont.	cont.	cont.	Cont.	term.	Cont.	cont.	Term.	cont.	Cont.
NbC_{1-x}	term.	cont.	Cont.	Cont.	cont.	term.	cont.	Cont.	term.	Cont.	cont.	cont.	term.	Cont.
TaC_{1-x}	term.	cont.	Cont.	Cont.	cont.	term.	cont.	Cont.	term.	Term.	cont.	cont.	term.	Cont.
ZrC_{1-x}	term.	cont.	Cont.	Cont.	cont.	Cont.	cont.	Cont.	cont.	Cont.	Cont.	term.	cont.	Cont.
						Oxides								
	EuO	MgO	NbO	(PuO)[f]	$TiO_{1\pm x}$	(UO)[f]	–	–	–	–	–	–	–	–
HfC_{1-x}	term.	term.	term.	term.	term.	term.	–	–	–	–	–	–	–	–
NbC_{1-x}	term.	term.	term.	term.	term.	term.	–	–	–	–	–	–	–	–
TaC_{1-x}	term.	term.	term.	term.	term.	term.	–	–	–	–	–	–	–	–
ZrC_{1-x}	term.	term.	term.	term.	term.	term.	–	–	–	–	–	–	–	–

(continued)

Table A.10 (continued)

Materials formula	Refractory compounds (with cubic NaCl structure)[a]													
	Phosphides													
	$CeP_{1\pm x}$	$LaP_{1\pm x}$	$PuP_{1\pm x}$	$ThP_{1\pm x}$	$UP_{1\pm x}$	$ZrP_{1\pm x}$	–	–	–	–	–	–	–	–
HfC_{1-x}	term.	term.	term.	term.	term.	term.	–	–	–	–	–	–	–	–
NbC_{1-x}	term.	term.	term.	term.	term.	term.	–	–	–	–	–	–	–	–
TaC_{1-x}	term.	term.	term.	term.	term.	term.	–	–	–	–	–	–	–	–
ZrC_{1-x}	term.	term.	term.	term.	term.	term.	–	–	–	–	–	–	–	–
	Sulphides													
	$CeS_{1\pm x}$	$DyS_{1\pm x}$	$ErS_{1\pm x}$	$EuS_{1\pm x}$	$GdS_{1\pm x}$	$HoS_{1\pm x}$	LaS_{1-x}	$NdS_{1\pm x}$	$PrS_{1\pm x}$	$PuS_{1\pm x}$	$SmS_{1\pm x}$	$ThS_{1\pm x}$	$US_{1\pm x}$	$YS_{1\pm x}$
HfC_{1-x}	term.	term.	term.	term.	term.	term.	term.	term.	term.	term.	term.	term.	term.	term.
NbC_{1-x}	term.	term.	term.	term.	term.	term.	term.	term.	term.	term.	term.	term.	term.	term.
TaC_{1-x}	term.	term.	term.	term.	term.	term.	term.	term.	term.	term.	term.	term.	term.	term.
ZrC_{1-x}	term.	term.	term.	term.	term.	term.	term.	term.	term.	term.	term.	term.	term.	term.

[a]The accepted abbreviations: Cont. – continuous solid solutions (complete solid solubility in the system, on the basis of published experimental data), cont. – the probable formation of continuous solid solutions (complete solid solubility in the system, on the basis of theoretical crystallochemical approach), Term. – terminal mutual solubilities (on the basis of published experimental data), term. – terminal mutual solubilities (on the basis of theoretical crystallochemical approach)

[b]At temperatures >1700–2000 °C

[c]In the temperature range of 2500–2850 °C

[d]In the temperature range of 2000–2600 °C

[e]In the temperature range of 2500–2800 °C

[f]The "imaginary" phases (no existence in the individual form)

Table A.11 General formulae for the relationships between the various physical properties (thermal conductivity, electrical resistance, strength and elastic characteristics) and porosity of the materials prepared by powder metallurgy methods (sintering, hot pressing, spark plasma sintering and others)

Mathematical expression	Designations	Formula number	Authors
	Thermal conductivity, λ[a]		
$\lambda = \lambda_0(1 - P)/(1 + 0.5P)$ (universal formula, for matrix systems)	λ_0 – thermal conductivity of poreless materials, P – volume fraction of porosity	(A.1)	Maxwell and Eucken
$\lambda = \lambda_0(1 - P)$ (applied for refractory carbides, nitrides and oxides)	" "	(A.2)	Kingery and Loeb
$\lambda = \lambda_0(1 - 1.2P)$ (applied for refractory carbides and oxides)	" "	(A.3)	Moore and McElroy
$\lambda = \lambda_0(1 - 1.5P)$ (universal formula, for statistical mixture, $0 < P \leq 2/3$)	" "	(A.4)	Odelevskii
$\lambda = \lambda_0(1 - P)/(1 + P)$ (universal formula, for lengthened pores)	" "	(A.5)	Odelevskii
$\lambda = \lambda_0(1 - P^{2/3})$ (universal formula)	" "	(A.6)	Kämpf and Karsten
$\lambda = \lambda_0(1 - P^{2/3})/(1 - P^{2/3} + P)$ (universal formula)	" "	(A.7)	Kämpf and Karsten
$\lambda = \lambda_0(1 - P)^{3/2}$ (universal formula, for isolated spherical pores)	" "	(A.8)	Balshin
$\lambda = \lambda_0(1 - P)^2$ (universal formula, for interconnected pores)	" "	(A.9)	Balshin

(continued)

Table A.11 (continued)

Mathematical expression	Designations	Formula number	Authors
$\lambda = \lambda_0(1 - P)^{5/3}$ (universal formula, for cylindrical pores)	" "	(A.10)	Ondracek et al.
$\lambda = \lambda_0 a(1 - P)$ (proposed for highly porous materials)	a – proportionality constant	(A.11)	Ashby
$\lambda = \lambda_0 \xi(1 - 1.5P)$ (universal formula, for sintered solids with imperfect contacts)	$\xi \approx r/R$, where r – welding (sintered) neck radius and R – particle radius	(A.12)	Skorokhod
$\lambda = \lambda_0(1 - P)^n$ with $n = a/2b + (1 - a)/(1 - b)$ (universal formula, for isolated closed pores with various forms)	a and b – stereometric parameters: orientation and form factors, respectively	(A.13)	Ondracek et al.
$\lambda = \lambda_0(1 - P)/(1 + aP^2)$ (applied for high-temperature materials)	a – constant (for δ-$TiN_{1\pm x}$: $a = 6$; for polycrystalline graphite and α-Al_2O_3: $a = 10$)	(A.14)	Aivazov and Domashnev
$\lambda = \lambda_0(1 - aP + bP^2)$ (applied for refractory metals and compounds)	a and b – constants (for metallic W: $a = 1.56$, $b = 0.311$; for ZrC_{1-x} ($0 < x \leq 0.35$) and NbC_{1-x} ($0.02 \leq x \leq 0.30$): $a = 0.3$ and $b = -5.5$)	(A.15)	Kulcinski et al.
	Electrical resistivity, ρ		
$\rho = \rho_0(1 + 0.5P)/(1 - P)$ (universal formula)	ρ_0 – electrical resistance of poreless materials	(A.16)	Maxwell
$\rho = \rho_0(1 + P)/(1 - P)$ (universal formula, for lengthened pores)	" "	(A.17)	Odelevskii
$\rho = \rho_0(1 + aP^2)/(1 - P)$ (applied for high-temperature materials)	a – constant (for polycrystalline graphite: $a = 8$)	(A.18)	Aivazov and Domashnev
$\rho = \rho_0(1 + aP + bP^2)$ (applied for refractory metals)	a and b – constants (for metallic W: $a = 1.83$, $b = 5.37$)	(A.19)	Kulcinski et al.

(continued)

Table A.11 (continued)

Mathematical expression	Designations	Formula number	Authors
	Strength characteristics, σ		
$\sigma = \sigma_0 \exp(-aP)$ (universal formula, for $P \leq 0.5$)	σ_0 – strength of poreless materials, a – constant (without a clear correlation to the porosity structure)	(A.20)	Ryshkewitch and Duckworth
$\sigma = \sigma_0(1 - aP)$ (universal formula)	a – constant (determined experimentally, without a clear correlation to the porosity structure; for sintered $ZrC_{0.99}$: $a = 2.8$ (for flexural strength) and 3.3 (for compressive strength))	(A.21)	Hasselman
$\sigma = \sigma_0(1 - P)^n$ (universal formula, proposed for sintered metallic materials)	n – exponential constant ($3 \leq n \leq 10$)	(A.22)	Balshin
$\sigma = \sigma_0 a(1 - P)^n$ (proposed for highly porous materials)	a – proportionality constant dependent on the pore geometry and n – exponential constant dependent on the pore morphology (for interconnected porosity $n = 2$)	(A.23)	Ashby
$\sigma = \sigma_0\{1 + aP/[1 - (a + 1)P]\}$ (proposed and applied for refractory materials)	a – constant (determined experimentally)	(A.24)	Hasselman
$\sigma = \sigma_0(1 - P)/(1 + aP)$ (proposed and applied for refractory materials)	a – constant (determined experimentally)	(A.25)	Weil
$\sigma = \sigma_0[1 - 1.5P(1 + 2a) + 4.5aP^2]$ (proposed for sintered metallic materials)	a – attenuation coefficient	(A.26)	Pines
$\sigma = \sigma_0\{(1 - A)^3/[A^3 + (1 - A)^3]\}$ with $A = (3P/4\pi)^{1/3} = r/l$ (proposed for materials with equal-sized pores)	r – radius of pores and l – distance between them	(A.27)	Harvey

(continued)

Table A.11 (continued)

Mathematical expression	Designations	Formula number	Authors
$\sigma = \sigma_0(1 - P)/(1 - 6.25P + aP/2^m\pi)^{1/m}$ (proposed for materials with low porosity ($P < 0.1$) and spherical pores)	a – constant and m – Weibull modulus	(A.28)	Peras and Dauknis
$\sigma = \sigma_0\{1 - [1.5abP(6P/\pi b - 36P/\pi^2)^{1/2}]\}$ (proposed for sintered metallic materials) or $\sigma = \sigma_0[(1 - 1.5P)/(1 + 1.5a'P)]$ (in an approximate form for the practical application)	a – coefficient of the inhomogeneity of pores distribution within a cross-section, b – coefficient of the variation of porosity (for tension $b > 1$ and for compression $b < 1$) and a – parameter that determines the inhomogeneity of stresses within a cross-section (for most of sintered materials $a' \approx 2$)	(A.29)	Troshchenko
$\sigma = \sigma_0(1 - P^2)^2\exp(-aP)$ (proposed for sintered metallic materials)	a – coefficient determined by the conditions of preparation and testing of the porous materials	(A.30)	Shcherban
$\sigma = \sigma_0[E(1 - P)/E_0]^{1/2}$ (proposed for sintered ceramic materials)	E_0 – Young's modulus of poreless materials	(A.31)	Carniglia
$\sigma = \sigma_0[1 - (P/P_0)^n]$ (applied for the evaluation of compressive strength of gypsum based materials)	n – exponential constant ($n = 3$ for spherical pores and $n = 2$ for cylindrical pores) and P_0 – percolation limit of solid (the porosity value corresponding to $\sigma = 0$)	(A.32)	Schiller
$\sigma = \sigma_0[1 - (P/P_0)^n][1 + a(P/P_0)^p]$ (applied for the evaluation of compressive strength of gypsum based materials)	a, n and p – empirically determined constants	(A.33)	Millard
$\sigma = 3\sigma_0(\xi^2/k)(1 - P)^{2/3}(1 - P_0)$ (applied for ultra-high temperature oxide materials produced from microspheres)	k – the coefficient of concentration of stresses in welding ligament formed due to sintering and $\xi \approx r/R$, where r – welding (sintered) neck radius and R – particle radius	(A.34)	Krasulin et al.

(continued)

Table A.11 (continued)

Mathematical expression	Designations	Formula number	Authors
	Young's modulus, E		
$E = E_0(1 - P)^3$ (proposed for sintered metallic materials)	E_0 – Young's modulus of poreless materials	(A.35)	Balshin
$E = E_0(1 - P)^n$ (based on porous body modelling approach and applied for refractory materials)	n – exponential constant ($0.5 \leq n \leq 4$, dependent on the tortuosity of the porous structure; for hot-pressed SiC: $n = 3.80$; for Si_3N_4 based materials: $n = 2.6$–5.5; for UO_{2+x}: $n = 2.27$; for RE oxide materials: $n = 2.0$–2.5, α-Al_2O_3: $n = 2.14$; β-Al_2O_3: $n = 4.12$)	(A.36)	Wagh et al.
$E = E_0 a(1 - P)^n$ (proposed for highly porous materials)	a – proportionality constant dependent on the pore geometry and n – exponential constant dependent on the pore morphology (for cellular materials with open pores (cells) $n = 2$ and with closed pores (cells) $n = 3$)	(A.37)	Ashby
$E = E_0 \exp(-aP)$ (universal formula, for $P \leq 0.5$)	a – constant ($1.4 \leq a \leq 9.0$, determined experimentally, without a clear correlation to the porosity structure; for β-SiC: $a = 2.73$; for $B_{4\pm x}C$: $a = 5.46 \pm 0.30$; for AlN: $a = 2.44$; for α-HfO_{2-x}: $a = 4.17$; for MgO: $a = 4.75$; for UO_{2+x}: $a = 2.51$; for α-Al_2O_3: $a = 1.60$–4.35)	(A.38)	Spriggs

(continued)

Table A.11 (continued)

Mathematical expression	Designations	Formula number	Authors
$E = E_0(1 - aP)$ (universal formula)	a – constant (determined experimentally, without a clear correlation to the porosity structure; for $UC_{1\pm x}$: a = 1.9–2.3; for UN_{1-x}: a = 2.0–2.7; for UO_{2+x}: a = 2.28; for α/β-Hf(Y)O_{2-x}: a = 2.60; for β-Hf(Er)O_{2-x}: a = 2.17; for MgO: a = 2.7–4.9; for RE oxide materials: a = 2.4–2.5; for glass materials: a = 2.06)	(A.39)	Hasselman
$E = E_0\exp[-a\,(1 - P)]$ (universal formula, for $P \geq 0.5$)	a – constant (determined experimentally, without a clear correlation to the porosity structure)	(A.40)	Rice
$E = E_0(1 - aP)/(1 + bP)$ (universal formula)	a and b – constants (determined experimentally, without a clear correlation to the porosity structure)	(A.41)	Hasselman
$E = E_0\exp[-(aP + bP^2 + cP^3 + \cdots)]$ (based on a model for packing equal-sized spheres) or $E = E_0\exp[-(aP + bP^2)]$ (in an approximate form applied for refractory materials)	$a, b, c\ldots$ – non-negative constants (additional high-order terms can be included in the exponential polynomial for wider porosity ranges and for improved accuracy; for α-Al_2O_3 ($0.05 \leq P \leq 0.32$): a = 1.46 and b = 9.82)	(A.42)	Wang
$E = E_0(1 + aP + bP^2)/(1 + cP)$ (based on the dependence of sound velocity on the elastic properties of materials)	a, b and c – constants dependent on the shape and size of the average pore and the properties of poreless materials	(A.43)	Kupkova

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Table A.11 (continued)

Mathematical expression	Designations	Formula number	Authors
$E = E_0(1 - aP)^n$ (based on porous body modelling approach and applied for refractory materials)	a and n – semi-empirical constants: packing geometry factor and grain morphology/pore geometry parameter, respectively (for SiC: a = 1.0 and n = 3.8; for Si_3N_4 based materials: a = 1.0–2.3 and n = 1.1–2.6; for UO_{2+x}: a = 1.0 and n = 4.1; for α-Al_2O_3: a = 1.0–3.9 and n = 0.7–3.4; for β-Al_2O_3: a = 1.0 and n = 4.1; for RE oxide materials: a = 1.9–2.7 and n = 0.7–1.3)	(A.44)	Phani and Niyogi
$E = E_0(1 - aP^n)$ (universal formula, for spherical pores, $P \leq 0.5$)	a = 1.21 and n = 2/3	(A.45)	Ondracek
$E = E_0\{1 - \pi b(9aP^2/16\pi^2)^{1/3}[1 + (1/a^2 - 1)]^{1/2}\}$ (proposed for the pores of various shapes in isotropic materials)	$a = z/x$ – the shape factor, defined by the axial ratio of the substitutional spheroids, and $b = \cos^2\alpha$ – the orientation factor, where α is the angle between the stress direction and the rotational axis of the substitutional spheroid (for isotropically oriented porosity: a = 4.5 and b = 0.33)	(A.46)	Ondracek
$E = E_0(1 - P^{2/3})^{1.21\Gamma}$ with $\Gamma = a^{1/3}[1 + b(a^{-2} - 1)]^{1/2}$ (proposed for the pores of various shapes in isotropic materials)	" "	(A.47)	Boccaccini and Ondracek
$E = E_0(1 - P^{2/3})/(1 + aP^n)$ (proposed for materials prepared from refractory compounds (carbides, nitrides, borides and silicides) by powder metallurgy methods)	a and n – empirical constants taking into account the effect of concentration of stresses by pores	(A.48)	Frantsevich
$E = E_0(1 - P^{2/3})/(1 + P - P^{2/3})$ (based on physico-mechanical analysis for a matrix-inclusion model)	–	(A.49)	Paul

(continued)

Table A.11 (continued)

Mathematical expression	Designations	Formula number	Authors
$E = E_0(1 - aP + bP^2)$ (based on a model for packing equal-sized spheres and applied for ultra-high temperature carbide and oxide materials)	a and b – constants (without a clear correlation to the porosity structure, to be determined experimentally; for carbide materials: $a \approx 1.9$–2.0 and $b \approx 0.9$–1.0, e.g. for ZrC_{1-x}: $a = 1.9$ and $b = 0.9$; for ThO_{2-x}: $a = 2.3$–2.7; for Y_2O_{3-x}: $a = 1.5$ and $b = -2.7$)	(A.50)	Knudsen et. al
$E = E_0\{1 + aP/[1 - (a + 1)P]\}$ (proposed and applied for refractory materials)	a – constant (determined experimentally; for BeO: $a = -2.0$–2.8)	(A.51)	Hasselman
$E = E_0 a(1 - P)^2/[P + a(1 - P)]$ (proposed and applied for refractory materials)	a – constant (parameter of structure, $0.4 \leq a \leq 1.0$)	(A.52)	Boccaccini and Fan
$E = E_0\{1 - 15P(1 - v_0)/[(7 - 5v_0) + 2P(4 - 5v_0)]\}$ (based on physico-mechanical approach, applied for ultra-high temperature carbide and carbide-carbon materials)	v_0 – Poisson's ratio of poreless materials	(A.53)	Hashin
$E = E_0[1 - 15P(1 - v_0)/(7 - 5v_0)]$ (based on physico-mechanical approach with the assumption of independence of Poisson's ratio on porosity, applied for ultra-high temperature carbide materials)	" "	(A.54)	Plyatt et al.
$E = E_0[1 - 3P(9 + 5v_0)(1 - v_0)/2(7 - 5v_0)]$ (based on physico-mechanical approach with the assumption of independence of Poisson's ratio on porosity)	" "	(A.55)	Hill
$E = E_0(1 - P)/[1 + (1 + v_0)(13 - 15v_0)/2P(7 - 5v_0)]$ (based on physico-mechanical approach and applied for ceramic materials)	" "	(A.56)	Weil
$E = E_0(1 - P)^2/[1 + (1 - 5v_0)(3v_0 - 1)P/2(7 - 5v_0)]$ (proposed for an isolated spherical pore geometry)	" "	(A.57)	Nielsen

(continued)

Table A.11 (continued)

Mathematical expression	Designations	Formula number	Authors
$E = E_0(1 - P)^2/[1 + P(2 - 3v_0)]$ (based on numerical experiments using the finite element method)	" "	(A.58)	Ramakrishnan and Arunachalam
$E = 3E_0\exp[-aP/(1 - P)]/\{2(1 + v_0)\exp[-bP/(1 - P)] + (1 - v_0)\}$ (applied for ultra-high temperature carbide materials)	a and b – constants, which express the influence of the shape, size distribution and relative orientation of the pores on the stress conditions (for $ZrC_{0.91}$: $a = 3.26$ and $b = 1.06$; for $ZrC_{0.95}$: $a = 3.49$ and $b = 1.08$; for $NbC_{0.98}$: $a = 2.30$ and $b = 1.18$)	(A.59)	Ordanyan et al.
	Coulomb's (shear) modulus, G		
$G = G_0\exp(-aP)$ (universal formula, applied for sintered ultra-high temperature oxide materials, for $P \leq 0.5$)	G_0 – Coulomb's (shear) modulus of poreless materials, a – constant (determined experimentally, without a clear correlation to the porosity structure; for MgO $a = 3.90$; for α-Al_2O_3: $a = 1.7$–3.3)	(A.60)	Spriggs
$G = G_0(1 - aP)$ (universal formula)	a – constant (determined experimentally, without a clear correlation to the porosity structure; for $UC_{1\pm x}$ and UN_{1-x}: $a = 1.92$; for α/β-Hf(Y)O_{2-x}: $a = 2.60$; for β-Hf(Er)O_{2-x}: $a = 2.15$; for glass materials: $a = 1.94$)	(A.61)	Hasselman
$G = G_0(1 - aP)/(1 + bP)$ (universal formula)	a and b – constants (determined experimentally, without a clear correlation to the porosity structure)	(A.62)	Hasselman

(continued)

Table A.11 (continued)

Mathematical expression	Designations	Formula number	Authors
$G = G_0\exp[-aP/(1-P)]$ (applied for ultra-high temperature carbide materials)	a – constant, which express the influence of the shape, size distribution and relative orientation of the pores on the stress conditions (for $ZrC_{0.91}$: $a = 2.20$; for $ZrC_{0.95}$: $a = 2.41$; for $NbC_{0.98}$: $a = 2.12$)	(A.63)	Ordanyan et al.
$G = G_0(1-aP)^n$ (based on porous body modelling approach and applied for refractory materials)	a and n – semi-empirical constants: packing geometry factor and grain grain morphology / pore geometry parameter, respectively (for Si_3N_4: $a = 1.0$ and $n = 2.9$)	(A.64)	Phani and Niyogi
$G = G_0(1-5P/3)$ (based on physico-mechanical approach)	–	(A.65)	Dewey
$G = G_0(1-aP+bP^2)$ (applied for ultra-high temperature oxide materials)	a and b – constants (without a clear correlation to the porosity structure, to be determined experimentally; for ThO_{2-x}: $a = 2.5$–2.9)	(A.66)	Knudsen et. al
$G = G_0\{1-aP/[1-(a+1)P]\}$ (applied for refractory materials)	a – constant (determined experimentally; for BeO: $a = 2.0$–2.8)	(A.67)	Hasselman
$G = G_0[1-5P(3K_0+4G_0)/(9K_0+8G_0)+AP^2]$ (for spherical pores, applied for sintered high-temperature materials)	K_0 – bulk (compression) modulus of poreless materials, A – proportionality coefficient for higher powers of porosity, determined by setting $G = 0$ at $P = 1$	(A.68)	MacKenzie
$G = G_0[1-5P(3K+G)/(9K+8G)]$ (universal formula)	–	(A.69)	Kerner
$G = G_0(1-P)/[1+2(4-5v_0)/P(7-5v_0)]$ (based on physico-mechanical approach and applied for ceramic materials)	v_0 – Poisson's ratio of poreless materials	(A.70)	Weil

(continued)

Table A.11 (continued)

Mathematical expression	Designations	Formula number	Authors
$G = G_0[1 - 15P(1 - v_0)/(7 - 5v_0)]$ (based on physico-mechanical approach with the assumption of independence of Poisson's ratio on porosity)	" "	(A.71)	Hill
$G = G_0(1 - P)^2/[1 + (1 - 5v_0)P/(7 - 5v_0)]$ (proposed for an isolated spherical pore geometry)	" "	(A.72)	Nielsen
$G = G_0(1 - P)^2/[1 + P(11 - 19v_0)/4(1 + v_0)]$ (based on numerical experiments using the finite element method)	" "	(A.73)	Ramakrishnan and Arunachalam
	Bulk (compression) modulus, K		
$K = K_0(1 - aP)$ (universal formula)	K_0 – bulk (compression) modulus of poreless materials, a – constant (determined experimentally, without a clear correlation to the porosity structure; for $UC_{1\pm x}$: $a = 2.52$; for UN_{1-x}: $a = 2.51$)	(A.74)	Hasselman
$K = K_0(1 - aP)/(1 + bP)$ (universal formula)	a and b – constants (determined experimentally, without a clear correlation to the porosity structure)	(A.75)	Hasselman
$K = K_0\{1 - aP/[1 - (a + 1)P]\}$ (applied for refractory materials)	a – constant (determined experimentally)	(A.76)	Hasselman
$K = K_0\exp[-aP/(1 - P)]$ (applied for ultra-high temperature carbide materials)	a – constant, which express the influence of the shape, size distribution and relative orientation of the pores on the stress conditions (for $ZrC_{0.91}$: $a = 3.26$; for $ZrC_{0.95}$: $a = 3.49$; for $NbC_{0.98}$: $a = 2.30$)	(A.77)	Ordanyan et al.
$K = K_0[1 - P(1 + 3K/4G)]$ (universal formula)	–	(A.78)	Kerner

(continued)

Table A.11 (continued)

Mathematical expression	Designations	Formula number	Authors
$K = K_0\{1 - 3P(1 - v_0)/[2(1 - 2v_0) + P(1 + v_0)]\}$ (based on physico-mechanical approach)	v_0 – Poisson's ratio of poreless materials	(A.79)	Hashin
$K = K_0(1 - P)/[1 + (1 + v_0)/2P(1 - 2v_0)]$ (based on physico-mechanical approach and applied for ceramic materials)	" "	(A.80)	Weil
$K = K_0[1 - 15P(1 - v_0)/(7 - 5v_0)]$ (based on physico-mechanical approach with the assumption of independence of Poisson's ratio on porosity)	" "	(A.81)	Plyatt et al.
$K = K_0(1 - P)^2/[1 + P(5v_0 - 1)/2(1 - 2v_0)]$ (proposed for an isolated spherical pore geometry)	" "	(A.82)	Nielsen
$K = K_0(1 - P)^2/[1 + P(1 + v_0)/2(1 - 2v_0)]$ (based on numerical experiments using the finite element method)	" "	(A.83)	Ramakrishnan and Arunachalam
$K = K_0(1 - P)/[1 + PG_0(1 + v_0)/2G(1 - 2v_0)]$ (based on the self-consistent method of mechanics of composites)	" "	(A.84)	Hill and Budiansky
$K = 2K_0G(1 - 2v_0)/G_0(1 + v_0)[1 + (G/G_0)^{3/5}(1 - 5v_0)/(1 + v_0)]$ (based on the differential method of mechanics of composites)	" "	(A.85)	Zimmerman
$K = 2K_0(1 - 2v_0)(3 - 5P)(1 - P)/[2(3 - 5P)(1 - 2v_0) + 3P(1 + v_0)]$ (proposed for the materials with low concentration of spherical pores)	" "	(A.86)	Ondracek
$K = 2(1 - s)K_0(1 - 2v_0)(3 - 5P)(1 - P)/[2(3 - 5P)(1 - 2v_0) + 3P(1 + v_0)] + 2sK_0(1 - 2v_0)(1 - P)/3(1 - v_0)$ with $s = 1/\{1 + \exp[-100(P - 0.4)]\}$ (proposed for the materials with spherical pores for the whole porosity range)	" "	(A.87)	Boccaccini and Ondracek
$K = 1/[1/K_0(1 - P) + 3P/4G_0(1 - P) + AP^3]$ (proposed for spherical pores, applied for sintered high-temperature materials)	A – proportionality coefficient for the higher powers of porosity	(A.88)	MacKenzie

(continued)

Table A.11 (continued)

Mathematical expression	Designations	Formula number	Authors
	Poisson's ratio, v		
$v = v_0 - aP$ (universal formula, applied for sintered ceramic materials, for $P \leq 0.5$)	v_0 – Poisson's ratio of poreless materials, a – constant (determined experimentally, without a clear correlation to the porosity structure; for $UC_{1\pm x}$: $a = 0.17$–0.29; for UN_{1-x}: $a = 0.10$–0.37; for α-Al_2O_3: $a = 0.30$–0.35)	(A.89)	Spriggs
$v = [2v_0(7 - 5v_0) + P(1 - 5v_0)(3 - v_0)]/[2(7 - 5v_0) + P(1 - 5v_0) \times (3v_0 - 1)]$ (proposed for an isolated spherical pore geometry)	–	(A.90)	Nielsen
$v = (4v_0 + 3P - 7v_0P)/4(1 + 2P - 3v_0P)$ (based on numerical experiments using the finite element method)	–	(A.91)	Ramakrishnan and Arunachalam
$v = 0.5 - (1 - P^{2/3})^{1.21}[2(3 - 5P)(1 - 2v_0) + 3P(1 + v_0)]/4(3 - 5P) \times (1 - P)$ (proposed for spherical pores in isotropic materials)	–	(A.92)	Boccaccini and Ondracek
$v = 0.5 - (1 - P^{2/3})^{1.21}/4\{(1 - s)(3 - 5P)(1 - P)/[2(3 - 5P) \times (1 - 2v_0) + 3P(1 + v_0)] + s(1 - P)/3(1 - v_0)\}$ with $s = 1/\{1 + \exp[-100(P - 0.4)]\}$ (proposed for the isotropic materials with spherical pores for the whole porosity range)	–	(A.93)	Boccaccini and Ondracek

[a]Most of formulae in this section can be applied for the description of resistivity as well

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