

# Chapter 10

## Structural Intermetallics

R. Mitra and R.J.H. Wanhill

**Abstract** Development of materials for structural applications at elevated temperatures in aeroengines has encouraged research on intermetallic alloys. A select group of aluminides and silicides has shown significant promise for high temperature structural applications owing to their high melting temperatures, as well as their ability to retain strength and oxidation resistance at elevated temperatures. In recent years the focus is on multiphase multicomponent intermetallic alloys with significant volume fractions of ductile constituents to achieve an optimum combination of toughness and elevated temperature strength. The engineering properties and actual or potential aerospace applications of the currently most important structural intermetallics, the nickel, iron, and titanium aluminides, are concisely discussed.

**Keywords** Intermetallics • Nickel aluminides • Titanium aluminides • Iron aluminides • Molybdenum silicides • Niobium silicides • Processing • Properties • Applications

### 10.1 Introduction

Intermetallics with high melting points are of interest for structural applications in aeroengine and automotive components, and also nuclear and power generation equipment requiring strength retention at elevated temperatures. Intermetallics are either compounds with fixed stoichiometric ratios between the metals, or phases having an extended range of compositions that may vary with temperature.

Stoichiometric intermetallics consist of strictly periodic arrangements of atoms. The formation and evolution of intermetallic phases in the microstructure depends

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on their thermodynamic stability. However, metastable phases with inhomogeneous compositions are often formed during solidification, and the equilibrium phases are subsequently obtained by a suitable heat treatment.

Much research has concentrated on development of intermetallics for various structural and functional applications. The properties of interest for high temperature aerospace applications are high melting point, strength retention, ductility, and resistance to environmental degradation.

This chapter provides a concise overview of the structural intermetallics based on aluminides and silicides. These have received the most attention over the past three decades owing to their promise for elevated temperature use.

For overviews and details of the fundamentals regarding phase equilibria, processing, structure, and properties of intermetallics as well as development of selected intermetallic types, the reader is referred to Refs. [1–7].

## 10.2 Crystal Structures and Compositions of Selected Intermetallics

The typical crystal structures, space-groups, and lattice constants of some of the structural intermetallics are shown in Table 10.1 [6].

**Table 10.1** Nominal compositions and crystal structures of selected intermetallics [6]

Composition	Crystal structure	Structure symbol and space group	Lattice constants (nm)
NiAl	Body-centered cubic	B2 (cP2), Pm-3m	0.2887
Ni <sub>3</sub> Al	Face-centered cubic	L1 <sub>2</sub> (cP4)	$a = 0.356$
Al <sub>3</sub> Ti	Tetragonal	D0 <sub>22</sub> (tI8), I4/mmm	$a = 0.3809$ ; $c = 0.847$
TiAl	Face-centered tetragonal	L1 <sub>0</sub> (tP4), P4/mmm	$a = 0.3997$ ; $c = 0.4081$
Ti <sub>3</sub> Al	Hexagonal	D0 <sub>19</sub> (hP8), P6 <sub>3</sub> /mmc	$a = 0.5729$ ; $c = 0.4574$
Fe <sub>3</sub> Al	Face-centered cubic	D0 <sub>3</sub> (cF16), Fm3m	0.5655
FeAl	Body-centered cubic	B2 (cP2), Pm-3m	0.291
MoSi <sub>2</sub>	Body-centered tetragonal	C11 <sub>b</sub> (tI6), I4/mmm	$a = 0.3202$ ; $c = 0.7845$
Mo <sub>5</sub> Si <sub>3</sub>	Body-centered tetragonal	D8 <sub>m</sub> (tI32), I4/mcm	$a = 0.959$ ; $c = 0.487$
Mo <sub>3</sub> Si	Cubic	A15 (cP8), Pm3n	0.4892
Mo <sub>5</sub> SiB <sub>2</sub>	Body-centered tetragonal	D8 <sub>1</sub> (tI32), I4/mcm	$a = 0.6013$ ; $c = 1.103$
Mo(Si,Al) <sub>2</sub>	Hexagonal	C40 (hP9), P6 <sub>2</sub> 22	$a = 4.644$ ; $c = 6.548$
NbSi <sub>2</sub>	Hexagonal	C40 (hP9), P6 <sub>2</sub> 22	$a = 4.7971$ ; $c = 6.592$
Nb <sub>5</sub> Si <sub>3</sub>	Body-centered tetragonal	$\alpha$ : D8 <sub>1</sub> (tI32) $\beta$ : D8 <sub>m</sub> , I4/mcm	$a = 0.656$ ; $b = 1.187$ $a = 1.0$ ; $b = 0.507$

### 10.2.1 Nickel Aluminides

The Ni–Al binary system contains the following intermetallic phases starting from the Al-rich side:  $\text{Al}_3\text{Ni}$ ,  $\text{Al}_3\text{Ni}_2$ ,  $\text{AlNi}$  (or  $\text{NiAl}$ ),  $\text{Al}_3\text{Ni}_5$ , and  $\text{AlNi}_3$  (or  $\text{Ni}_3\text{Al}$ ) [7, 8], see Fig. 10.1.  $\text{NiAl}$  and  $\text{Ni}_3\text{Al}$  are the most important, owing to their high melting/liquidus temperatures:

- $\text{NiAl}$  has a wide temperature-dependent homogeneous phase field ranging from 40–69 at.% Ni, with the highest congruent melting point (1638 °C) at the stoichiometric composition (50:50 at.%).
- $\text{Ni}_3\text{Al}$  has a homogeneity range of  $\approx 4.5$  at.% around its stoichiometric composition, and a melting point of 1395 °C at the stoichiometric composition.

Table 10.1 shows that  $\text{NiAl}$  has an ordered body-centered structure (B2, cP2), whereas  $\text{Ni}_3\text{Al}$  has an ordered face-centered cubic (fcc) structure ( $\text{L1}_2$ , cP4). The ordered structure in  $\text{NiAl}$  remains stable even above  $0.65T_M$  (melting point temperature in K). This is attributed to strong covalent bonding along  $\langle 111 \rangle$  directions between the nearest neighbour Ni–Al atom pairs [9]. This bonding is accompanied by weak ionic repulsion between second nearest neighbour atoms along  $\langle 100 \rangle$ .

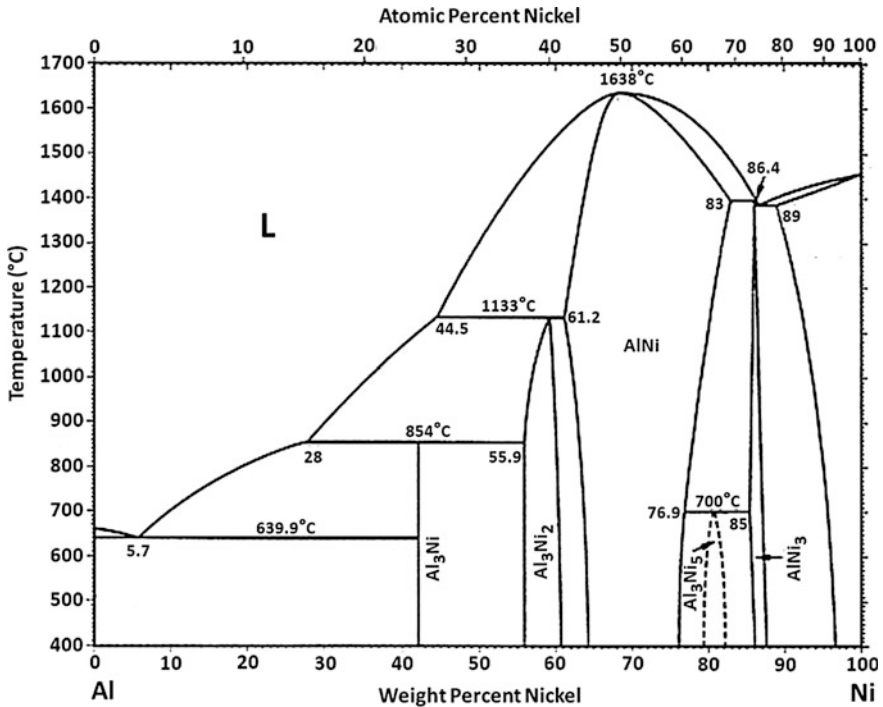


Fig. 10.1 Ni–Al phase diagram: after [7]

Both types of directional bonds are superimposed upon the nondirectional metallic bonds between unlike and similar metallic atoms and are responsible for the elastic anisotropy of NiAl.

### 10.2.2 Titanium Aluminides

The Ti–Al binary system contains three intermetallic phases,  $\text{Al}_3\text{Ti}$ ,  $\text{TiAl}$ , and  $\text{Ti}_3\text{Al}$  [10], with a posited  $\text{Ti}_2\text{Al}$  phase at 66 at.% Al, see Fig. 10.2. Only  $\text{TiAl}$  ( $\gamma$  TiAl alloys) and  $\text{Ti}_3\text{Al}$  ( $\alpha$ -2 alloys) are of engineering importance.

**Ti–Al:** The binary intermetallic exists over a wide range (49–66 at.% Al) of temperature-dependent stability, and apparently remains ordered up to its melting point of about 1450 °C [7].

The  $\gamma$  alloys of engineering importance contain about 45–48 at.% Al and 1–10 at.% M, where  $M = \text{V}, \text{Cr}, \text{Mn}, \text{Nb}, \text{Ta}, \text{or W}$ , or combinations of some of these elements. These  $\gamma$  alloys may be divided into two categories: single-phase ( $\gamma$ ) and two-phase ( $\gamma + \alpha$ -2).

The single-phase alloys contain additional alloying elements such as Nb or Ta to improve strength and oxidation resistance. Two-phase alloys have additions of V, Cr, and Mn to increase ductility; Nb and Ta for increased oxidation resistance; or combinations of these elements to improve both ductility and oxidation resistance [7]. The maximum temperature capability, above which protective coatings must be used, is about 750–800 °C, depending on long- or short-term applications, e.g. in gas turbines or missile fins [7].

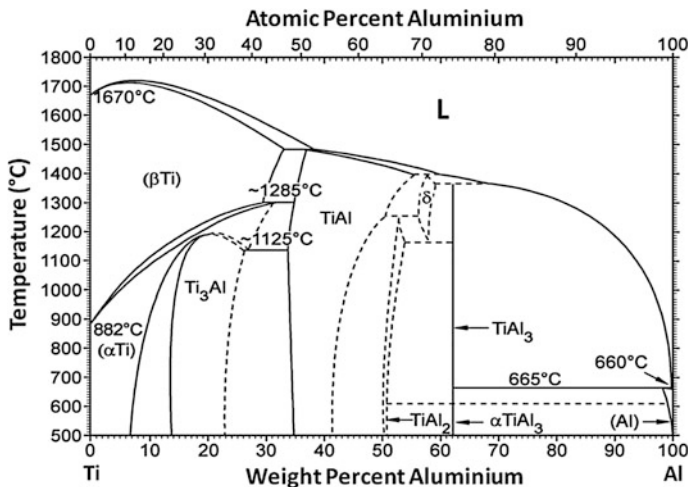


Fig. 10.2 Ti–Al phase diagram: after [7]

**Ti<sub>3</sub>Al:** As shown in Fig. 10.2, the  $\alpha$ -2 alloy has a wide range (22–39 at.%) of compositional stability. The stoichiometric composition (Ti-25 at.% Al) is stable up to about 1090 °C [7], while the limit of ordering in any composition is about 1180 °C [11].

The hexagonal crystal structure (see Table 10.1) is retained when small amounts of alloying additions (Nb, V, Mo) are made to improve creep resistance while retaining an Al content  $\approx$  25 at.% [7]. However, additions of Nb above about 10 at.% result in a change to an ordered orthorhombic structure based on the O phase (Ti<sub>2</sub>AlNb).

The incentives for higher alloying and changing the crystal structure are significant improvements in strength and ductility over a wide range of temperatures, though at the expense of maximum temperature capability (about 650 °C) [7].

These Ti–Al–Nb alloys have been studied for several decades and can be classified according to the amount of  $\beta$ -stabilizers (Nb, Mo, Ta and V) [11] as:

- (1) 10–12 at.% Nb. These alloys have two-phase microstructures consisting of  $\alpha$ -2 and  $\beta$  or B2 (disordered or ordered bcc phase).
- (2) 14–17 at.% Nb. These alloys contain three phases:  $\alpha$ -2,  $\beta$  (or B2), and O.
- (3) 25–30 at.%  $\beta$ -stabilizers. These alloys are reported to have two-phase microstructures consisting of  $\beta$  (or B2) and O.

### 10.2.3 Iron Aluminides

The Fe–Al binary system contains the following intermetallic phases starting from the Fe-rich side: Fe<sub>3</sub>Al, FeAl, FeAl<sub>2</sub>, Fe<sub>2</sub>Al<sub>3</sub>, Fe<sub>2</sub>Al<sub>5</sub>, and FeAl<sub>3</sub>. [7, 12], see Fig. 10.3. Only Fe<sub>3</sub>Al and FeAl are of interest for structural applications, since the rest are very brittle.

**Fe<sub>3</sub>Al:** This intermetallic has an ordered fcc D0<sub>3</sub> structure up to 540 °C, and then an ordered body-centered cubic B2 structure up to 760 °C [7]. Above this temperature the B2 structure becomes disordered.

The D0<sub>3</sub>  $\rightarrow$  B2 transition temperature decreases with increasing Al content above 25 at.%, and only the B2 structure is stable at Al contents above 36 at.% [7]. The most commonly used composition is Fe-28 at.% Al.

**FeAl:** This ordered B2 phase has a wide range of compositional stability, approximately 30–50 at.% Al, as may be seen from Fig. 10.3. The most commonly used composition is Fe-40 at.% Al.

### 10.2.4 Molybdenum Silicides

The Mo–Si binary phase diagram shows the presence of two stoichiometric compounds with compositions Mo<sub>3</sub>Si and MoSi<sub>2</sub>; and the compound Mo<sub>5</sub>Si<sub>3</sub>, which has

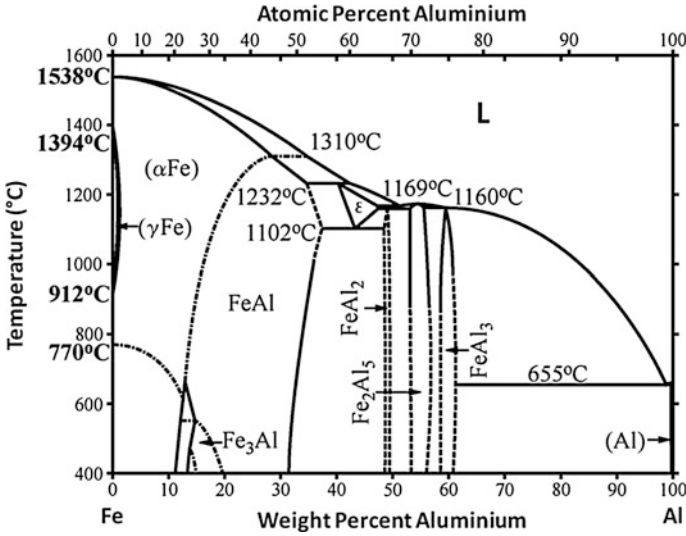


Fig. 10.3 Fe–Al phase diagram: after [7]

a homogeneity range of 3 at.% Si [13], see Fig. 10.4. Most interest for practical applications has been directed to MoSi<sub>2</sub> itself and Mo–Si–B alloys containing Mo, Mo<sub>3</sub>Si, and Mo<sub>5</sub>SiB<sub>2</sub>.

**MoSi<sub>2</sub>:** This has an ordered body-centered tetragonal structure up to 1900 °C. It is attractive because of its high temperature capabilities, but is very hard and brittle at ambient temperatures.

R&D on this compound for structural applications has been directed toward rather exotic high temperature composite materials with MoSi<sub>2</sub> as matrix, and ceramic or refractory metal reinforcements [7].

**Mo–Si–B alloys:** These have potential for advanced gas turbine engines with turbine entry temperatures (TET) beyond the capabilities of nickel-base superalloys [14].

As an example, Fig. 10.5 shows the Mo-rich section of the ternary isothermal phase diagram [14] of the Mo–Si–B system at 1600 °C, indicating the various phases. The ternary alloys can be designed to have the optimum volume fractions of Mo<sub>ss</sub> (Mo-solid solution), Mo<sub>3</sub>Si, and Mo<sub>5</sub>SiB<sub>2</sub> phases. All three phases have nearly fixed composition with limited solubility for other elements, and they therefore provide microstructural stability at high temperatures.

### 10.2.5 Niobium Silicides

The Nb–Si binary phase diagram shows the presence of two intermetallic compounds with compositions Nb<sub>5</sub>Si<sub>3</sub> and NbSi<sub>2</sub>, see Fig. 10.6. Nb<sub>5</sub>Si<sub>3</sub> has a very

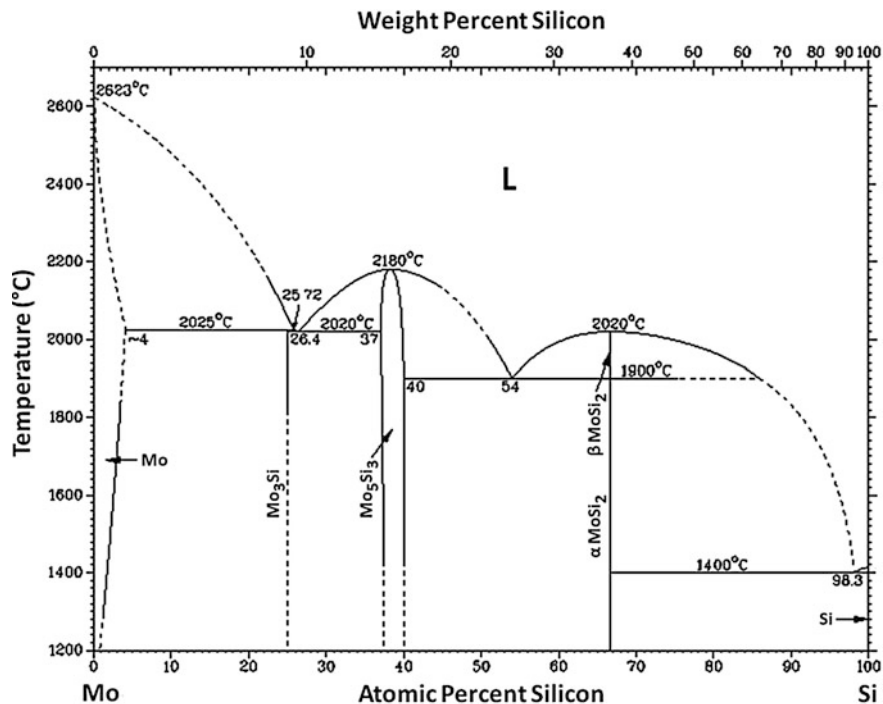
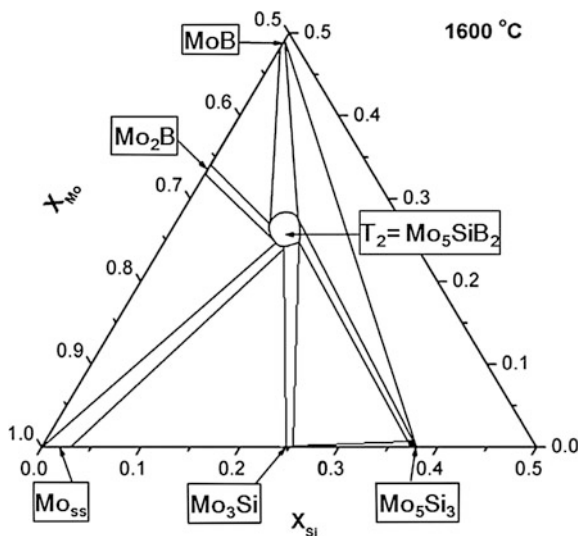


Fig. 10.4 Mo-Si phase diagram: after [13]

Fig. 10.5 Mo-Si-B isothermal section for Mo-rich compositions for Mo-rich compositions at 1600 °C [14]



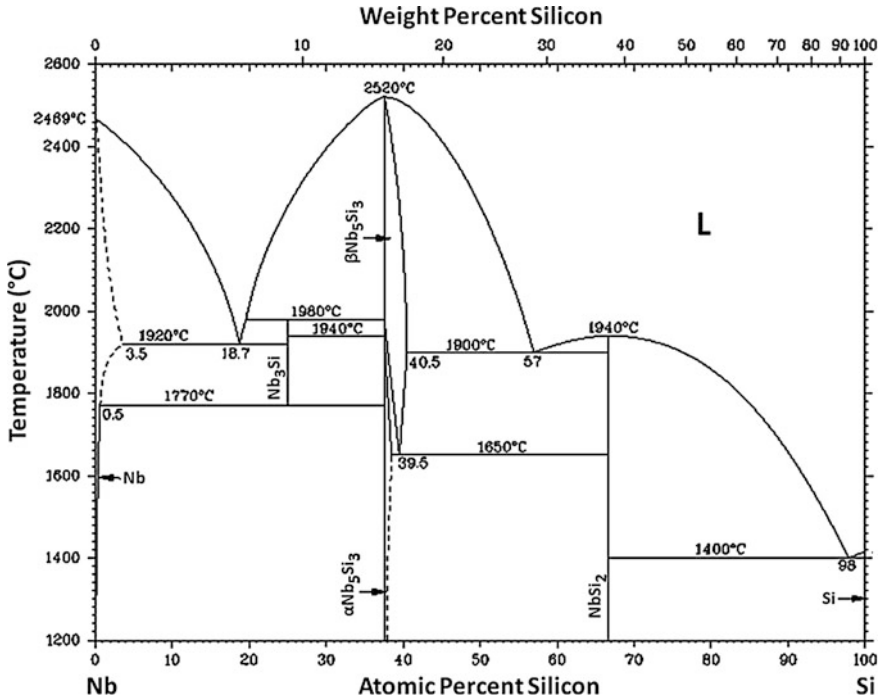


Fig. 10.6 Nb–Si phase diagram: after [15]

small homogeneity range at lower temperatures ( $\alpha\text{Nb}_5\text{Si}_3$ ) and a homogeneity range of about 3 at.% Si when changed to  $\beta\text{Nb}_5\text{Si}_3$ . A eutectoid reaction at around 1770 °C leads to decomposition of the  $\text{Nb}_3\text{Si}$  phase into Nb solid solution ( $\text{Nb}_{\text{ss}}$ ) and  $\text{Nb}_5\text{Si}_3$ .

As in the case of molybdenum silicide alloys, the niobium silicide alloys have potential to replace nickel-base superalloys in advanced gas turbines. Since the monolithic phase  $\text{Nb}_5\text{Si}_3$  is very brittle at ambient temperatures, multiphase composites are envisaged based on  $\text{Nb}_5\text{Si}_3$  alloyed with other elements (e.g. Ti) and employing a ductile niobium solid solution ( $\text{Nb}_{\text{ss}}$ ) as the matrix [16]. This is ongoing R&D, with actual applications still in the future.

### 10.3 Processing

**General information:** Bulk intermetallics can be processed by conventional ingot metallurgy (IM) and powder metallurgy (PM) routes, as well as by rapid solidification processing (RSP). Coatings are prepared by various methods including electron beam physical vapour deposition (EB-PVD), which has become



increasingly favoured for aerospace-quality coatings on gas turbine and other components. The coatings are generally of the NiCoCrAlY (oxidation resistant) or CoCrAlY (hot corrosion resistant) types.

The purity of raw materials and the processing route control the final impurity content, and also influence the microstructural evolution, thereby influencing the properties. Special methods of processing have been developed to control the impurity levels, the grain sizes, and the sizes of dispersed phases, all of which affect the mechanical properties

**IM processing:** The ingot metallurgy processing methods include arc-melting [17], induction skull melting [18], single crystal growth [19], directional solidification [20], and investment casting [21].

There are also several special techniques to produce in situ composites [22]. These are not limited to intermetallics and are currently more in the developmental stage than ready for actual production.

**PM processing:** The powder metallurgy processing methods involve consolidation of the powders to form densified intermetallic alloys of a specified composition. The powders can have either the specified composition of the intermetallic alloy, or be synthesized reactively from elemental precursors.

For some intermetallic alloys reactive synthesis of elemental powders is combined with uniaxial hot pressing or hot isostatic pressing (HIP). A reaction is started by applying heat in the form of a laser pulse or electric discharge to one surface of the elemental powder compact [23]. The reactions during synthesis are exothermic and release sufficient heat to drive them to completion. If the reaction initiation temperature is higher than the melting point of one of the reactant elements, then the presence of a transient liquid phase enhances the reaction kinetics and the process of densification, as occurs for nickel aluminides.

**RSP methods:** These include both the formation of powders for PM processing and thermal spray deposition on a substrate. Thermal spray deposition has some applications for bulk products and coatings, e.g. the ‘Osprey’ process [24], but the quality is insufficient for aerospace bulk products and also coatings on critical components such as high pressure turbine blades and vanes.

**Secondary fabrication:** Nickel, iron, and titanium aluminides can be fabricated by IM and PM, and also as castings [7, 25]. Iron and titanium aluminide ingots can be conventionally hot-rolled, but nickel aluminide ingots must be isothermally forged since there is only a narrow temperature window before the ingot surfaces cool to become brittle and crack [7]. However, conventional (secondary) hot forging is feasible for fine-grained nickel aluminides [7].

PM consolidation can be done, e.g. by steel-canning and hot forging or extrusion, and also by HIP followed by forging or extrusion.

## 10.4 Properties of Ni-, Fe-, and Ti-Based Aluminides

Nickel, iron, and titanium aluminides are currently the most likely intermetallics to (partly) replace nickel-base superalloys in high-performance aeroengines. The main obstacles to their use have been, and continue to be, brittleness and low ductility at ambient temperatures [7].

Much R&D has been directed to enhance the low temperature ductility, and also improving the high temperature properties. This includes the development of in situ composites, mentioned in Sect. 10.3.

The properties of interest for high temperature aerospace applications are high specific strength and stiffness (favoured by low densities), high melting points, strength retention and creep resistance, ductility, and resistance to environmental degradation.

### 10.4.1 Property Surveys

**Mechanical properties:** Table 10.2 lists some generic properties and characteristics of nickel, iron, and titanium aluminides. Table 10.3 compares the properties of titanium aluminides with those of titanium alloys and nickel-base superalloys. Table 10.4 lists the mechanical properties of molybdenum and niobium silicides. From these tables the following points can be made:

1. The titanium aluminides offer large advantages in density compared to nickel-base superalloys, and are also less dense than conventional titanium alloys. TiAl is especially attractive because of its high elastic modulus (and hence high specific stiffness), and high creep and oxidation resistance.

**Table 10.2** Some properties and characteristics of nickel, iron, and titanium aluminides [7]

Alloy	Melting point (°C)	Density (g/cm <sup>3</sup> )	R.T. E (GPa)	Maximum use (°C)		Alloy characteristics
				Strength	Corrosion	
Ni <sub>3</sub> Al	1390	7.50	179	1000	1150	Oxidation resistance; high temperature strength
NiAl	1640	5.86	294	1200	1400	Oxidation resistance; high melting point
Fe <sub>3</sub> Al	1540	6.72	141	600	1100	Oxidation and sulphidation resistance
FeAl	1250	5.56	261	800	1200	Oxidation and sulphidation resistance
Ti <sub>3</sub> Al	1600	4.2	145	760	650	Low density; good specific strength
TiAl	1460	3.91	176	1000	900	Low density; good specific strength
MoSi <sub>2</sub>	2020	6.24	440			

**Table 10.3** Properties of titanium aluminides, titanium alloys, and nickel-base superalloys [7]

Alloys	Density (g/cm <sup>3</sup> )	R.T. E (GPa)	R.T. Y.S. (MPa)	R.T. UTS (MPa)	Creep limit (°C)	Oxidation limit (°C)	Ductility (%)	
							R.T.	H.T.
Ti-alloys	4.5	96–100	380–1150	480–1200	600 <sup>a</sup>	600	20	High
<b>Ti<sub>3</sub>Al</b>	<b>4.1–4.7</b>	<b>100–145</b>	<b>700–990</b>	<b>800–1140</b>	<b>760</b>	<b>650</b>	<b>2–10</b>	<b>10–20</b>
<b>TiAl</b>	<b>3.7–3.9</b>	<b>160–176</b>	<b>400–650</b>	<b>450–800</b>	<b>1000</b>	<b>900<sup>b</sup></b>	<b>1–4</b>	<b>10–60</b>
Ni-alloys	8.3	206	–	–	1090 <sup>a</sup>	1090 <sup>a</sup>	3–5	10–20

<sup>a</sup>1995 data; <sup>b</sup>Needs coating above 750–800 °C, see Sect. 10.2.2

**Table 10.4** Properties of Mo-silicide and Nb-silicide based binary and multicomponent intermetallic alloys [5, 6]

Alloys	Melting point (°C)	Density (g/cm <sup>3</sup> )	R.T. E (GPa)	R.T. Flexural strength (MPa)	R.T. Fracture toughness (MPa√m)	Creep limit (°C)	Oxidation limit (°C)
MoSi <sub>2</sub>	2020	6.24	440	140–160	2.5–5.0	1300	1700
Mo <sub>3</sub> Si	2025	8.9	295	–	3.0	1300	1000
Mo <sub>5</sub> SiB <sub>2</sub>	2160–2200	8.8	383	–	2.0	1500	1400
Mo-Si-B (Mo <sub>88</sub> -Mo <sub>3</sub> Si-Mo <sub>5</sub> SiB <sub>2</sub> )	2025	8–9	310–325	400–600	5.0–20.0	1400	1300
Nb <sub>88</sub> -Nb <sub>5</sub> Si <sub>3</sub>	≥1915	7–7.5	170	500–1450	6–24	1200	1200
Nb-Si multicomponent alloys	≥1800	6.6–7.2	165	800–950	18.2–23.3	1200	1300

2. The nickel aluminides have lower densities than nickel-base superalloys, notably NiAl, which also has a much higher elastic modulus.
3. The iron aluminides also have lower densities than nickel-base superalloys and are resistant to high temperature oxidation and corrosion. However, the maximum strength capability of the iron aluminides is relatively low.
4. The silicides and multicomponent silicide-based alloys have much higher melting points and higher temperature capabilities compared to aluminides, but have higher densities. Furthermore, these materials are brittle at ambient temperatures in polycrystalline form, with the brittle-to-ductile temperature being very high, at about 1100 °C.

While useful, Tables 10.2, 10.3, and 10.4 are *generic*. They cannot provide the whole story, since (i) they do not fully cover the ambient and elevated temperature properties, and (ii) they do not account for the large amount of R&D on alloy processing and development over the past three decades. Both of these topics are beyond the necessarily limited scope of the present chapter. The reader is therefore referred, as mentioned earlier, to Refs. [1–7].

Nevertheless, it is important to note that the characteristic low ductility or brittleness of these intermetallics at ambient temperatures is (or has been) a major disadvantage with respect to (i) alloy processing and component manufacture, and (ii) service performance.

The first problem has been the focus of much fundamental and applied research to improve the ductility and toughness through alloying additions and process-based microstructural control. For example, Ref. [7] lists some 13 compositional variations and 19 processing methods for TiAluminides. An analogous, though less extensive, situation exists for Ni<sub>3</sub>Al [7], whose commercial alloys contain small additions of boron (B) to enhance the ambient temperature ductility [7, 26]; and varying amounts of Cr, Fe, Mo, and Zr [7].

There has also been much R&D directed to improving the high temperature properties, particularly the creep resistance of nickel aluminides to enable them to compete with advanced superalloys. However, although detailed comparisons are not available, the latest generations of nickel-base superalloys (discussed in Chap. 9 of this Volume of the Source Books) would appear to have high temperature properties beyond the capabilities of nickel aluminides, which have a temperature limit  $\approx 1100$  °C [27].

**Oxidation:** The nickel aluminides have an inherently high oxidation resistance, especially NiAl, which is why this alloy is used in bondcoats for thermal barrier coatings (TBCs). The high oxidation resistance is due to the formation of a protective Al<sub>2</sub>O<sub>3</sub> scale.

The oxidation behaviour of the iron aluminides is determined by the nature and composition of the oxide scale. The minimum concentration of Al to enable a continuous and protective Al<sub>2</sub>O<sub>3</sub> scale is reported to be about 15 at.% Al in a binary Fe-Al alloy [28]. Addition of Cr accelerates the formation of this protective scale [29].

The titanium aluminides TiAl and Ti<sub>3</sub>Al do not form a protective scale. However, certain alloying additions (Mo, Nb, Si) improve the oxidation resistance by enabling the formation of a continuous Al<sub>2</sub>O<sub>3</sub> scale [30].

Much more information on the oxidation characteristics of all these aluminides (and the silicides) is available in the literature, e.g. Ref. [2].

## 10.5 Aerospace Applications

### 10.5.1 Silicides

Earlier in this chapter it was stated that molybdenum and niobium silicides are potential replacements for nickel-base superalloys in advanced gas turbines. When this major change will occur, if it occurs, is unknown. For example, a decade ago, in 2006, General Electric (GE) suggested that niobium silicide low-pressure turbine (LPT) blades could be in use by 2012, with high pressure turbine (HPT) blades

following by 2015 [31]. This suggestion has been overtaken by events (continued superalloy improvements), but it does show the direction of thinking by a major engine manufacturer.

## 10.5.2 *Aluminides*

**Nickel and iron aluminides:** The prospects for using nickel and iron aluminides in aerospace gas turbines are uncertain. As mentioned in subsection 10.4.1, nickel aluminides may not be competitive with the latest generation of superalloys, even though NiAl has a significantly lower density (compare the data in Tables 10.2 and 10.3). In 1996 Darolia et al. [32] summarized the problems thus:

1. Nickel-base superalloys have properties that are extremely difficult to surpass.
2. Any intermetallics intended to replace superalloys will probably retain vestiges of one or more of the following generic problems:
  - Low ductility and toughness, especially at low temperatures.
  - Obtaining a balance of sufficient creep strength, ductility, and toughness.
  - Environmental stability.

Darolia et al. [32] concluded that the use of intermetallics will be very selective and gradual, and it is unlikely that superalloys will be rapidly and widely displaced. This conclusion is reinforced by superalloy developments and improvements in the subsequent two decades, see Chap. 9 of this Volume of the Source Books.

Furthermore, there seems no reason why iron aluminides should show any benefits in the relatively non-corrosive environments pertaining in aeroengine gas turbines.

**Titanium aluminides:** TiAl alloys are the first aluminides to have reached operational status in aeroengines, as 6th and 7th stage cast blades for the LPTs of the General Electric GENx-1B and GENx-2B engines, see Figs. 10.7 and 10.8. Similar components will be used in the GE9x engine, which is currently (early 2016) still under development.

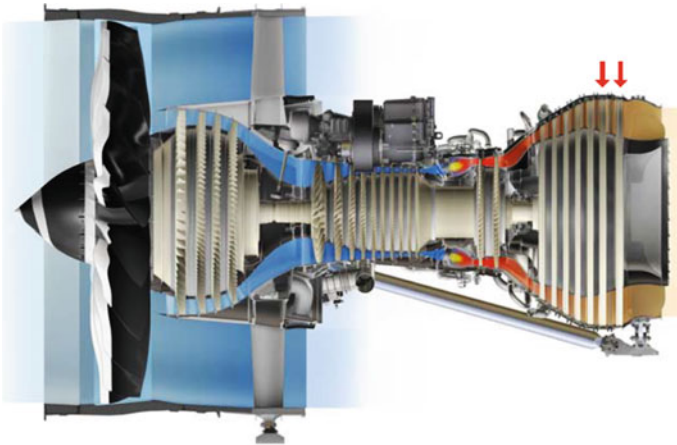
The GENx engines entered service in late 2011 with the Boeing 787 *Dreamliner* (GENx-1B), and the Boeing 747-8F (GENx-2B). Testing had already demonstrated that TiAl blades with redesigned (thicker) leading edges should survive relatively large impacts from foreign object ingestion [33]: this is important for *all* engine blades.

Nevertheless, it is noteworthy that this first application of TiAl blades is in the last stages of the engines. Thus any service failures would cause minimal damage. As service experience with TiAl LPT blades accumulates, it may be possible to consider aluminide blades for replacing conventional titanium alloys in some of the last stages of the high pressure compressor (HPC).

Other, potential, aeroengine applications of titanium aluminides include compressor casings and blade dampers; and critical components for the exhaust nozzle



**Fig. 10.7** TiAl LPT turbine blade for the GENx aeroengine. *Source* GE



**Fig. 10.8** LPT turbine blade locations (*arrowed*) in the GENx aeroengine: modified Boeing illustration

of an (eventual) future high speed civil transport (HSCT) [34]. However, it should be noted that ceramic matrix composites (CMCs) could be serious contenders [31].

### 10.5.3 Indian Scenario

Hindustan Aeronautics Limited (HAL) in Bengaluru has developed the precision forging of Ti<sub>3</sub>Al aluminide compressor blades, the alloy supplier being Mishra Dhatu Nigam (MIDHANI) in Hyderabad. Also, the Defense Metallurgical Research Laboratory in Hyderabad has demonstrated isothermal forging of in-house developed TiAl alloys [35].

These activities were carried out as part of the indigenous Kaveri engine program, which, however, was discontinued in 2015. Some information on the *structural* design of the Kaveri engine is given in Chap. 24 of Volume 2 of these Source Books.

## 10.6 Summary

Intermetallics are a class of materials that have been studied extensively, motivated by a variety of potential structural applications in the aerospace, automotive, and power generation industries. The materials discussed in this chapter are (i) nickel, iron, and titanium aluminides, and (ii) molybdenum and niobium silicides. These have been, or are, candidates for applications in aerospace gas turbines.

Only one intermetallic, TiAl and its alloying variants, has attained service use in aircraft engines. It is uncertain or unlikely that nickel and iron aluminides will do so. The main reason is that conventional nickel-base superalloys already have outstanding combinations of properties and are continuously being improved.

Molybdenum and niobium silicide composites are potential candidates for advanced engines with turbine entry temperatures beyond the capabilities of superalloys. However, they will have to compete with other materials such as ceramic matrix composites.

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