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INTERNAL NITRIDING OF HIGH-TEMPERATURE STEELS AND ALLOYS

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INTRODUCTION

The possibility of using nitriding for improving the service characteristics of high-temperature materials was first shown in the study of nitriding processes in refractory metals and alloys [1, 2]. It was established that the high-temperature strength could be increased by processes of internal nitriding,[2] which yield diffusion layers without a surface nitrided zone. This is accompanied by the formation of a zone of internal nitriding that consists of disperse particles of thermodynamically stable nitrides distributed in the matrix, which stabilize the structure at a high temperature, providing an effective dispersion hardening.

It is known that the service temperatures of dispersion-hardening steels and alloys are restricted by the conditions of stability of carbide and intermetallic particles and do not exceed 800 – 950°C [3].

Analysis of the thermodynamic characteristics of such phases (for example, the free energy of their formation [4]) has shown that many nitrides possess a higher thermodynamic stability than intermetallic compounds and carbides. This opens up possibilities for creating high-temperature alloys with nitride reinforcement provided by nitriding, i.e., saturation of the ready part with nitrogen from a gas phase.

The main scientific problem of the study of nitriding of high-temperature steels and alloys is the necessity of taking into account the behavior of all the alloying elements in the nitriding process, including their mutual interaction, the effect of the dissolution and diffusion of nitrogen in the alloy, and the susceptibility to nitride formation.

Internal nitriding can be treated as dispersion hardening by nitrides with simultaneous solid-solution hardening of the matrix by nitrogen. The hardening is the most effective if the disperse nitride formed in the matrix possesses a high thermodynamic stability.

Some of the conditions for successful occurrence of processes of internal nitriding of matrixes based on iron, nickel, and cobalt have been formulated in [5].

Nickel and cobalt have a positive value of the heat of nitride formation and hence do not form stable nitride phases under any actual nitriding conditions. The conditions of the process of internal nitriding for iron-based alloys can be chosen so as to eliminate the formation of iron nitrides. A high equilibrium nitrogen pressure above the matrix nitrides makes it possible to perform internal nitriding of Fe, Ni, and Co alloys at a partial pressure of nitrogen equal to the atmospheric one (0.1 MPa).

The nitride-forming elements in such alloys can be Ti, Zr, Hf, or Nb, which possess a high affinity for nitrogen.

The increase in the strength of the zone of internal nitriding depends, on one hand, on the nitrogen concentration in the solid solution and, on the other hand, on the nature, morphology, and distribution of nitride phases in the matrix [6]. The dispersity and distribution of the hardening nitride phases is a function of the following parameters: the concentration of the alloying element that forms the nitrides, the free energy of nitride formation, the diffusion mobility of the alloying element, the solubility of nitrogen in the matrix of the base metal, and the temperature and duration of the nitriding process. The properties of parts hardened by internal nitriding can be optimized by changing the chemical composition and the technical parameters of the nitriding process, namely, the temperature, the duration, and the value of the nitrogen potential of the atmosphere in different stages of the process. At the same time, the nitriding parameters should be chosen with allowance for the optimum geometry of the particles and the necessity of obtaining diffusion layers of the requisite thickness in a reasonable time. It should be noted that the dissolution and the diffusion mobility of nitrogen in alloys with an austenitic structure are low; therefore,

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[2] Despite the ambiguity and disputability of the term “internal nitriding,” we preserve it in the present paper, paying tribute to the school of Professor Yu. M. Lakhtin (*Ed. note*).

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TABLE 1. Chemical Composition of the Studied Alloys, Duration of Through Nitriding, and Microhardness of Nitrided Layers

Alloy		τ_{hr}, h	*HV*
Fe – 10 % Ni – 18 % Cr – 1,5 % Ti (steel Kh18N10T)	(*1*)	7	340
Ni – 25 % Cr – 10 % W – 15 % Co – 1,7 % Ti	(*2*)	23	445
Ni – 25 % Cr – 10 % W – 15 % Co – 2,1 % Ti	(*3*)	35	480
Ni – 25 % Cr – 10 % W – 1.7 % Ti	(*4*)	23	465
Ni – 33 % Cr – 10 % W – 1.8 % Ti	(*5*)	18	470
Ni – 25 % Cr – 10 % W – 15 % Co – 1 % Zr	(*6*)	15	320
Ni – 25 % Cr – 10 % W – 15 % Co – 1.7 % Zr	(*7*)	21	360
Co – 20 % Ni – 20 % Cr – 5 % Mo – 7 % W – 1.9 % Ti	(*8*)	25	420

Notes. 1. The alloys are numbered in parentheses for convenience.
2. The duration of through nitriding τ_{hr} was calculated for the following conditions: saturation in a nitrogen atmosphere at a temperature t_n = 1200°C.

for the saturation rate of such alloys to be high enough, the nitriding should be performed at a high temperature.

The aim of the present work consisted of developing a scientifically substantiated technology for internal nitriding of austenitic steels and alloys based on nickel and cobalt in order to raise their high-temperature strength and service temperature and work out recommendations on the technological parameters of nitriding for providing an optimum combination of the high-temperature characteristics in the materials (the high-temperature strength, the heat resistance, and the resistance to heat transfer).

METHODS OF STUDY

We performed the nitriding at a temperature ranging from 900 to 1200°C. The minimum duration of the process was 5 h. The maximum saturation time corresponded to through nitriding of specimens of the chosen alloy.

The saturating media were pure nitrogen, ammonia, or nitrogen–argon mixtures. This allowed us to study the effect of the nitrogen potential on the structure and properties of the nitrided alloys. The temperature of the decomposition of the nitride of the base metal in the studied alloys was much lower than that of the nitriding. Therefore, carbides of the base metal did not form upon saturation under atmospheric pressure.

We chose for our study austenitic steel Kh18N10T with different titanium contents (0.7 and 1.5%) and laboratory and commercial alloys based on nickel and cobalt and bearing various nitride-forming elements (Ti, Zr, Hf, Nb) in an amount ranging from 0.3 to 3%. We nitrided sheet specimens 1.5 mm thick after cold plastic deformation. The carbon content in the alloys was 0.005 – 0.04%. The chemical composition of some of the studied alloys is presented in Table 1.

We studied the microstructure of nitrided alloys on transverse microsections under a light metallographic microscope and a scanning electron microscope at different magnifications. The phase composition of the nitrided layers was evaluated with the help of x-ray diffraction analysis in a DRON-3 installation and by analyzing microscopic electron diffractograms taken from replicas extracted from the cross section of the nitrided specimens.

The microhardness of the nitrided specimens was measured using a PMT-3 device at 1-N load on the indenter. We determined the distribution of the microhardness over the thickness of the nitrided layer. The mechanical properties of the specimens were evaluated after nitriding and, for comparison, after quenching.

The tensile tests at room and elevated temperatures were performed for standard flat specimens 1.5 mm thick and 65 mm long in an R-5 rupture machine at a deformation rate of 2.5 mm/min (GOST 1497–84 and 9651–84).

The tests for the long-term strength were performed for standard flat specimens 1.5 mm thick and 65 mm long in an AIMA-5-2 testing machine (GOST 10145–81). We determined the time before failure of the specimens after testing them at specified temperatures and stresses.

The high-temperature strength was evaluated from the gain in the weight of the specimens after a 100-h test at 1100°C. We also determined the rate of gas corrosion. The tests for the heat resistance of the specimens were performed by subjecting them to thermocycling by the regime $20 \rightleftarrows 1100$°C. The time of the heating and the cooling was 1 min.

RESULTS AND THEIR DISCUSSION

Nitriding of multicomponent alloys should be performed with allowance for the fact that they can contain several nitride-forming elements. For example, both chromium nitrides and titanium nitrides can form under certain conditions in alloys containing chromium and titanium simultaneously. The formation of chromium nitrides is undesirable because they do not provide structural stability at a high temperature; when chromium is bound in nitrides, its content in the solid solution decreases, which diminishes the high-temperature strength and the corrosion resistance of the alloys. We wanted to determine nitriding conditions under which the segregation of chromium nitrides becomes thermodynamically impossible. For this purpose, we considered the effect of the chemical composition of the alloys and the nitriding parameters on the structure and phase composition of the nitrided layers.

Figure 1 presents a typical microstructure of a nitrided layer. The diffusion layers are well discernible on the microsections in the form of more etched regions that have a distinct boundary with the unnitrided core. The core of a plastically deformed specimen after nitriding has a coarsegrained structure with well-manifested twins, which allows us to speak of recrystallization.

After nitriding, the microstructure of most of the studied alloys did not have a surface zone of continuous nitrides, i.e., there occurred processes of internal nitriding. The absence of a nitride zone promoted the formation of thick zones of internal nitriding that consisted of disperse nitride particles distributed in the solid solution. The nature of these particles was determined by analyzing the electron diffractograms taken from extracted replicas. The predominant disperse phases in the zone of internal nitriding of the alloy bearing chromium and titanium were titanium nitride TiN (f.c.c. lattice with parameter $a = 0.454$ nm) and chromium nitrides CrN (f.c.c. lattice with $a = 0.414$ nm) and Cr_2N (hexagonal lattice, $a = 0.4759$ nm, $c = 0.4438$ nm).

The nitrided layers in the studied alloys were homogeneous (Fig. 1) and consisted of two regions. Directly at the surface we observed a relatively small dark region with segregations occurring predominantly over grain boundaries, and then followed an extended region with more or less uniformly distributed particles.

An x-ray diffraction analysis of the surface of the nitrided specimens showed the presence of CrN and Cr_2N nitrides. In some cases we identified a π-phase, i.e., a compound nitride whose composition depended on the chemical composition of the alloy. For example, in alloy *3* the composition of the π-phase was $(Cr, W)_{12}(Ni, Co)_8N_{4-y}$ (by the data of [7], the parameter of the cubic lattice of the π-phase $a = 0.6323$ nm).

In order to study the changes in the phase composition over the thickness of the zone of internal nitriding, we performed an x-ray diffraction analysis of various regions of the layer. For this purpose, we grinded-off layers of various thicknesses from the surface of the specimens. Analyzing the diffractograms taken from various regions of the layer on alloy *3* nitrided at 1100°C in nitrogen for 15 h, we established that the surface region mainly contained chromium nitrides and a π-phase. The major part of the zone of internal nitriding bore titanium nitrides (Fig. 1).

The thermodynamically caused formation of different nitrides in the zone of internal nitriding depends on the parameters of the process, i.e., the temperature, the hold time, the composition of the saturating medium that determines the nitrogen potential of the atmosphere, and the chemical composition of the nitrided alloys. In order to form nitrided layers with optimum properties, we should recommend such nitriding parameters that make it possible to eliminate the formation of chromium nitride on the surface of the alloy. This can be achieved by the following measures.

1. Nitriding should be performed at a high temperature (1150 – 1200°C) when the formation of chromium nitride is thermodynamically impossible. After nitriding in pure nitrogen at 1200°C, most alloys do not bear CrN and the π-phase; after nitriding at 1200°C for a long time (for over 20 h), Cr_2N nitrides disappear too. Thus, we can assume that at a partial pressure of nitrogen equal to the atmospheric one ($p = 0.1$ MPa), the limiting temperature of the existence of chromium nitrides is equal to 1200°C; for some alloys with a low Cr/Ti proportion, the temperature is lower (1150°C).

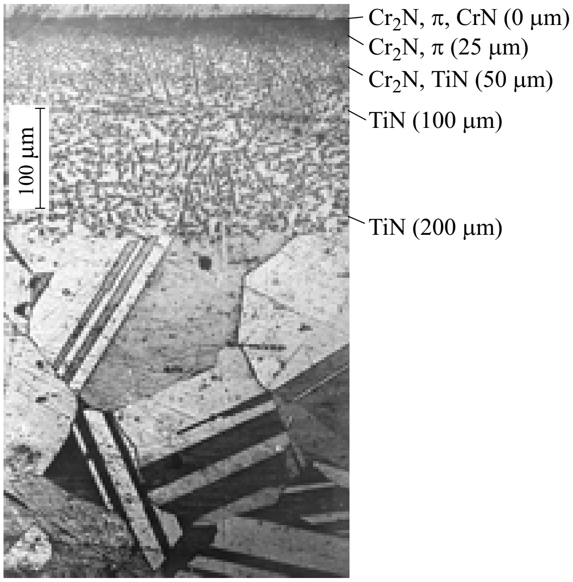

Fig. 1. Microstructure (× 250) and phase composition of a nitrided layer on alloy *3* after 15-h saturation at 1100°C in nitrogen. (The distance from the surface to the layer with the considered composition is given in parentheses.)

2. The saturating medium should have a diminished nitrogen potential (for example, nitrogen diluted with argon). The decrease in the nitrogen potential can change considerably the conditions of nitride formation, and the temperature of their maximum stability will be lower. For example, nitriding of a mixture of 15% N_2 + 85% Ar at 1150°C did not yield chromium nitrides and the π-phase in most cases, except for alloys with a high proportion of Cr/Ti, where the observed chromium nitride (Cr_2N) was the most stable. However, a decrease in the nitrogen potential and in the nitriding temperature increases considerably the duration of the saturation process requisite for obtaining layers with a great thickness.

3. The production process should consist of two stages involving nitriding at 1200°C in a medium of pure nitrogen and subsequent annealing for 1.5 – 2.5 h at the same temperature in a medium of an inert gas. In the annealing process in argon, the chromium nitrides formed in the nitriding stage decompose, the concentration of nitrogen in the solid solution increases, and the thickness of the zone of internal nitriding that consists of disperse titanium nitrides increases due to the diffusion processes.

4. The chemical composition of the alloys should be controlled by decreasing the amount of chromium and the strongest nitride-forming element.

Studies of the kinetics of nitriding processes have shown that the rate of growth of the zone of internal nitriding obeys

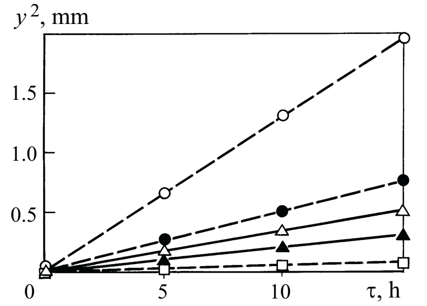

Fig. 2. Dependence of the thickness (y^2) of the zone of internal nitriding in steel Kh18N10T bearing 1.5% Ti (●, ▲) and 0.7% Ti (○, △) and in nickel-based alloy *3* (□) on the nitriding time: solid lines) $t_n = 1000°C$; dashed lines) $t_n = 1100°C$.

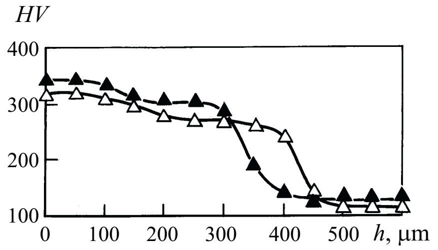

Fig. 3. Variation of the microhardness over the thickness of the layer in steel Kh18N10T bearing 0.7% Ti (△) and 1.5% Ti (▲) after 5-h nitriding at 1000°C (*h* is the distance from the surface).

a parabolic law. This shows that the process is controlled by diffusion. It can be seen from the data of Fig. 2 that as the titanium concentration in alloy Kh18N9T increases to 1.5%, the growth of the diffusion layer decelerates, especially at a high temperature. Nickel markedly decreases the dissolution of nitrogen in austenite, and therefore nitriding of nickel-based alloys occurs more slowly than in austenitic steels. With increase in the nitriding temperature the growth of the layer is strongly accelerated, which provides a zone of internal nitriding over the entire cross section of thin sheets.

We used experimentally obtained data with allowance for the parabolic dependence of the thickness of the layer on the nitriding time to calculate the hold time requisite for through nitriding of a specimen 1.5 mm thick over the entire cross section. Analyzing the data of Table 1, we see that nitriding occurs more rapidly in iron-based alloys than in alloys based on nickel and cobalt (compare, for example, alloys *1*, *2*, and *8*), but the level of hardening evaluated in terms of the microhardness in this case is lower.

The growth of the concentration of chromium in the alloys accelerates the nitriding process (alloys *4* and *5*) without considerable changes in the microhardness. The acceleration of the saturation is connected with the intensification of the dissolution of nitrogen in high-chromium alloys.

The increase in the content of the nitride-forming element (titanium or zirconium) decelerates the nitriding, but the microhardness of the layer increases (alloys *2*, *3* and *6*, *7*). It is shown in [6] that with an increase in the concentration of the nitride-forming element the volume fraction of the nitride particles increases, which is responsible for the higher microhardness.

The presence of cobalt in nickel alloys virtually does not affect the growth rate of the layer (alloys *2* and *4*).

Comparing the actions of various nitride-forming elements, we established that in alloys with zirconium the saturation occurs somewhat faster than in alloys with titanium (alloys *2* and *7*), but the microhardness of the nitrided alloys bearing titanium is substantially higher.

Analyzing the results of the study of the structure of zones of internal nitriding and the kinetics of the growth of the nitrided layers, we made preliminary conclusions on the preferable chemical compositions of high-temperature alloys hardened by nitriding. The zones of internal nitriding grow the most effectively in austenitic steels, and through layers reinforced by nitrides are formed in the minimum time. The hardening is the most considerable in nitriding of nickel-based alloys bearing chromium and titanium.

The formation of dispersed particles of the hardening nitride phase in the zone of internal nitriding increases the microhardness of the diffusion layer relative to the microhardness of the unnitrided core. We see from the characteristic profiles of the distribution of the microhardness over the thickness of specimens of steels of type Kh18N10T (Fig. 3) that the microhardness of the layer on the steel with an elevated content of titanium is higher and the thickness of the reinforced region is less. The curves describing the distribution of the microhardness over the thickness of the layer have two plateaus that correspond to the regions of segregation of the different nitride phases mentioned above. The near-surface region possesses a higher hardness than the subsequent layers, which is explainable by the elevated density of the segregations of chromium nitrides.

A longer duration of nitriding virtually does not affect the microhardness; only the thickness of the hardened layer increases. With increase in the temperature of nitriding, the level of the microhardness decreases somewhat, which is connected with coarsening of the nitride particles.

Mechanical tests of specimens of steel Kh18N10T have shown (Fig. 4) that after nitriding the ultimate rupture strength σ_r is 1.2 – 1.5 times higher than after quenching. The gain in the hardness is maximum after nitriding at 900°C; after nitriding at 1150 – 1200°C, the hardening effect is somewhat lower. As the nitrogen temperature t_n increases to 1050°C, the elongation δ of the steel decreases due to the increase in the thickness of the zone of internal nitriding. The minimum elongation detected at $t_n = 1050°C$ corresponds to the formation of a through zone of internal nitriding in a specimen 1.5 mm thick. In addition, the decrease in the ductility of the steel at this nitriding temperature is explainable by the formation of a high amount of chromium nitrides, segregated, as has been shown in [8], over grain boundaries,

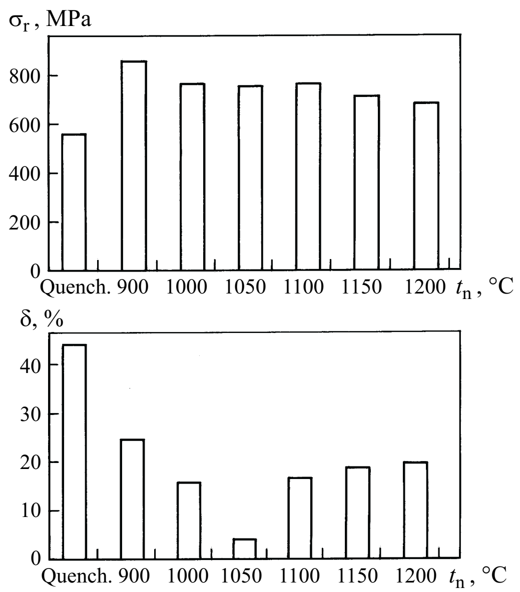

Fig. 4. Mechanical properties (σ_r and δ) of steel Kh18N10T after 15-h nitriding at different temperatures. Quench.) after air quenching from 1100°C.

which increases the susceptibility of the steel to brittle fracture. At higher nitriding temperatures, the amount of chromium nitrides decreases. At $t_n \geq 1150°C$, they virtually disappear due to the occurrence of the limit of their thermodynamic stability. As a result, the elongation increases to $\delta = 18\%$. The increase in the ductility with increase in the nitriding temperature above 1050°C is also connected with the coagulation of nitride particles in high-temperature nitriding. A some decrease in the elongation of the steel after nitriding with respect to the value of δ in the quenched state does not cause technological difficulties, because no additional treatment is needed after nitriding.

It can be seen from Fig. 5 that the extension of nitriding causes a smooth increase in the ultimate rupture strength of the alloys at 1100°C in correspondence with the growth of the thickness of the zone of internal nitriding. The elongation decreases even in the first hours of the saturation process, but for the nickel alloys it remains high enough (no less than 30%). At t_n = 1000°C, the thickness of the nitrided layer in the nickel-based alloy is low ($h_n \leq 80$ μm). Therefore, we do not observe noticeable hardening and decrease in the ductility. The higher the nitriding temperature, the greater the gain in the strength, which is connected with the more intense growth of the thickness of the nitrided layer. The upper limit of the nitriding temperature (1200°C) is determined by the conditions of coagulation of nitride particles.

The level of the high-temperature strength depends much on the amount of the nitride-forming element, which determines the volume fraction of the hardening nitride phase and the thickness of the zone of internal nitriding. The results of earlier studies made in [9] show that with increase in the content of the nitride-forming element the volume fraction of nitride particles increases and the thickness of the nitrided layer decreases. It can be seen from Fig. 6 that when the concentration of titanium in the nickel alloy subjected to short-term nitriding increases to 2%, its high-temperature strength increases and its high-temperature ductility decreases. Consequently, the determining factor in the given case is the increase in the volume fraction of nitrides. At a titanium concentration exceeding 3%, the strength of the nitrided layers decreases and the ductility increases again. This can be associated with the lower thickness of the zone of internal nitriding in alloys with an elevated concentration of titanium.

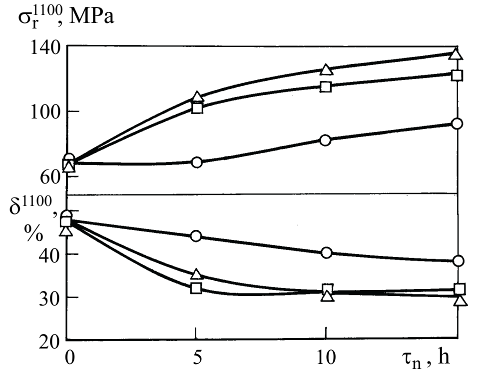

Fig. 5. Dependence of the mechanical properties σ_r^{1100} and δ^{1100} of alloy *2* on the nitriding time at the following temperatures: ○) t_n = 1000°C; □) t_n = 1100°C; △) t_n = 1200°C.

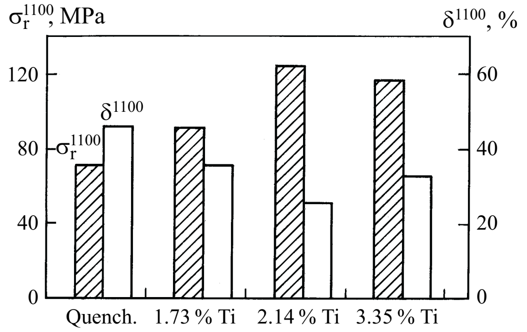

Fig. 6. Mechanical properties (σ_r^{1100} and δ^{1100}) of nickel-based alloys with different contents of titanium after 5-h nitriding at 1200°C in pure nitrogen (Quench. is used to denote quenching from 1200°C).

The results of short-term tensile tests of specimens of austenitic steel Kh18N10T are presented in Fig. 7. It can be seen that after the nitriding the short-term strength of the steel with 1.5% Ti at an elevated temperature is 25 – 40% higher than that of the steel with 0.7% Ti after nitriding or quenching. For example, the ultimate rupture strength of nitrided steel with 1.5% Ti at 900°C is higher than that of the unnitrided steel with 0.7% Ti at 800°C. The maximum service temperature of the unnitrided steel is 800°C. The service temperature of nitrided austenitic steels can be increased by 100 – 150°C for parts with short-term service life. The duc-

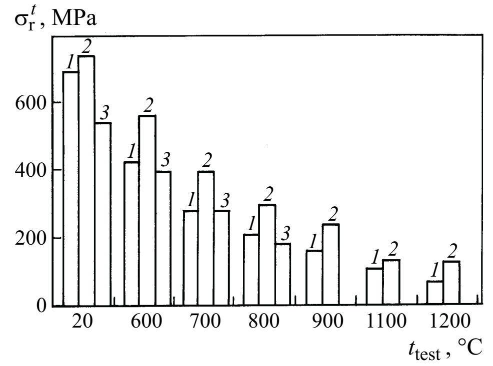

Fig. 7. Ultimate rupture strength σ_r^t at various temperatures t_{test} of steel Kh18N10T after nitriding at 1100°C in nitrogen (*1*, *2*) and after air quenching from 1100°C without nitriding (*3*): *1*, *3*) 0.7% titanium; *2*) 1.5% titanium.

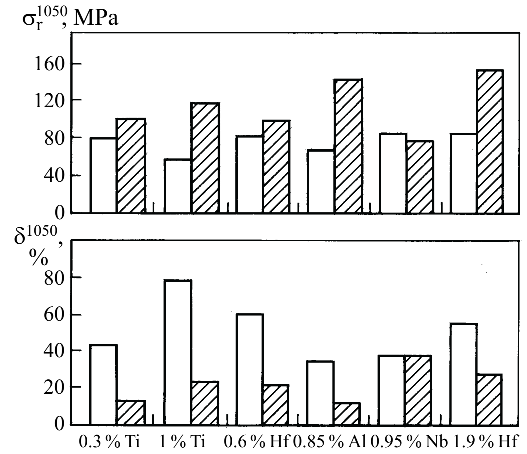

Fig. 8. High-temperature characteristics (σ_r^{1050} and δ^{1050}) of Ni – 20% Cr – 10% Co nickel alloys with additives of various nitride-forming elements: white columns) after quenching from 1200°C; hatched columns) after 14-h nitriding at 1200°C in an ammonium atmosphere.

tility of nitrided steels decreases but remains at an acceptable level.

The ultimate rupture strength at elevated temperatures is increased after nitriding of nickel alloys with titanium, hafnium, and aluminum. The mechanical properties of the alloys with niobium do not change (Fig. 8). On the whole, we established that the higher the thermodynamic stability of the nitride formed in the layer, the higher the high-temperature characteristics of the nitrided alloy. Though the elongation of the alloys decreases after nitriding, it remains at a level of $\delta \geq 20\%$, except for the alloy with aluminum.

Figure 9 presents the results of tests for the long-term strength at $t = 1100$°C and $\sigma = 40$ MPa for various nitrided alloys. With increase in the duration of the nitriding process, the thickness of the nitrided layer increases. Consequently, the increase in the time before failure is directly connected with the growth of the thickness of the hardened layer. After the specimen is nitrided throughout (Fig. 9), the time before failure decreases due to the development of coagulation of nitride particles. The elongation measured in the tests for long-term strength changes for different alloys from 25 to 40%.

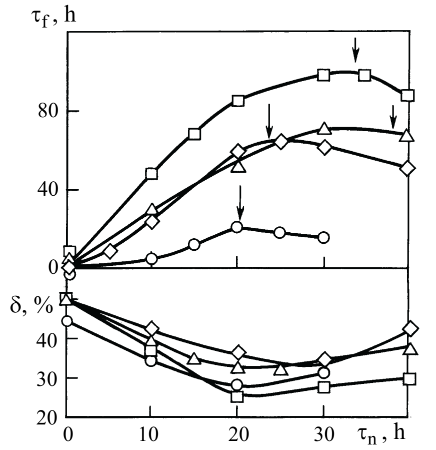

Fig. 9. Dependences of the time before failure τ_f and elongation δ in tests for long-term strength ($\sigma = 40$ MPa, $t = 1100$°C) on the duration of nitriding in nitrogen ($t_n = 1200$°C) for nickel alloys with additives of titanium and zirconium (the arrows indicate the design time of through nitriding of specimens 1.5 mm thick): ◇) 1.7% Ti; □) 2% Ti; △) 2.5% Ti; ○) 1.7% Zr.

The influence of the nitriding parameters (the temperature and duration of the process) on the characteristics of the high-temperature strength is determined considerably by the change in the thickness of the layer. The highest characteristics of the high-temperature strength correspond to the maximum nitriding temperature of 1200°C used in our experiment; at this temperature, the formed diffusion layers have maximum thickness at one and the same nitriding time.

The long-term strength of nitrided steels and alloys based on nickel and cobalt exceeds the initial value in all the cases. As a result of through nitriding of specimens of nickel alloy 1.5 mm thick with an optimum composition, their 100-h long-term strength at 1100°C increases by a factor of 3.5; that of the cobalt alloy increases by a factor of 2.8 relative to the initial value.

For parts serving in aggressive media at a high temperature, the heat resistance, i.e., the resistance to the gas corrosion at the service temperature, is as important as the high-temperature strength. Fracture of the surface layers of metals and alloys at a high temperature in a gas medium presents a serious problem in operation of machine parts. High-temperature nickel alloys and austenitic steels possess a quite high high-temperature strength at a high concentration of chromium in them. The oxide film on the surface of the alloys at a high temperature is enriched by chromium oxides of the Cr_2O_3 type. In this connection, the nitriding should not cause bonding of chromium in nitrides; the whole of the chromium should be in the solid solution.

Specimens of nickel alloy *3* were tested for high-temperature strength after 15-h nitriding at 1200°C.

Below we present the results of the evaluation of the gain in weight of specimens of nitrided and unnitrided alloy *3* and, for comparison, of specimens of some commercial alloys tested for high-temperature strength.

Alloy	Gain in weight, g/(m^2 · h)
3	0.375/0.369
ÉI437B	0.4
ÉI617	1.0

Notes. 1. The weight of the unnitrided specimen is given in the numerator; that after the nitriding is given in the denominator.

2. The depth of corrosion of nitrided alloy *3* is 0.0654 μm/h.

It can be seen that the high-temperature strength of the nitrided alloy evaluated in terms of the gain in weight of the specimens does not worsen relative to the initial value. A comparative analysis has shown that the nitrided alloy *3* possesses a quite high high-temperature strength at service temperatures (up to 1100°C) close to and even exceeding that of commercial alloys.

Many parts from high-temperature alloys are subjected to the action of heat cycles in operation. In this case, the internal stresses can be high due to the nonuniformity of the temperature gradient even without the action of external loads and can cause cracking and warping of the parts.

The heat resistance of the alloys is evaluated in terms of the number of heat cycles that cause the appearance of the first crack in the specimen. Table 2 presents the results of tests of nitrided and unnitrided specimens for heat resistance and the data on the heat resistance of alloy Kh20N80T tested under the same conditions. It can be seen that the heat resistance of the studied alloy after through nitriding increases by 30%.

TABLE 2. Thermal Stability of Alloy *5* and Kh20N80T

Alloy	State	N_{crack}, cycles
5	Initial	67
	After nitriding	96
Kh20N80T [3]	Without heat treatment	80
	Air quenching from 1150°C	20

Note. We present the average number of cycles before cracking N_{crack} for specimens of alloy *5* determined from the results of six tests for thermal stability.

CONCLUSIONS

1. From the standpoint of elevation of the high-temperature strength of alloys based on iron, nickel, and cobalt and alloyed by nitride-forming elements, the favorable structure of the zone of internal nitriding should be stabilized by disperse particles of refractory nitrides (TiN, ZrN, HfN, etc.) that preserve their stability up to 1200°C. This increases the service temperature of the high-temperature alloys.

2. A zone of internal nitriding with a maximum thickness in the absence of a zone of chemical compounds is created on the surface by performing nitriding at an elevated temperature. Optimum results have been obtained in through nitriding of thin-walled parts (up to 3 mm).

3. After high-temperature nitriding, the service characteristics of high-temperature alloys are improved, namely,

– the ultimate rupture strength of the alloys at service temperatures of 900 – 1100°C increases by a factor of 1.5 – 2 at a high ductility (δ is no lower than 18 – 25%);

– the time before failure of nitrided alloys serving at 1100 – 1200°C increases by one or two orders of magnitude relative to that of unnitrided alloys;

– the 100-h long-term strength at a service temperature of 1100°C increases by a factor of 2 – 3.5 for the alloy with an optimum composition;

– the service temperatures of austenitic steels increase by 100 – 150°C with respect to the standard heat treatment; for nickel-based alloys they increase by 200°C;

– the service characteristics of commercial high-temperature and heat-resistant alloys are enhanced (for example, the 100-h strength at 1100°C of nitrided alloy ÉP648 is 20 – 25 MPa, which is 1.5 – 2 times higher than in the same alloy with intermetallic hardening):

– nitrided alloys preserve a high enough high-temperature strength up to 1100°C, and their heat resistance (20 ⇄ 1100°C) increases by 30%.

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