



Structural High-Temperature and (β NiAl+ γ)-Alloys Based on Ni-Al-Co-Me Systems with an Improved Low-Temperature Ductility

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Summary:

The β NiAl-based alloys (B2) have lower density higher resistance to oxidation, and higher melting temperature relative to those of Ni-superalloys or γ Ni₃Al-base alloys. An improved low-temperature ductility of advanced Ni-Al-Co-M β + γ alloys (EI=9-16% at 293-1173 K is achieved due to the formation γ -Ni solid solution intergranular interlayers of eutectic origin. Secondary γ and/or γ' precipitates form in the grains of the supersaturated β -solid solution upon heat treatment at 1473-1573 K and 1073-1173 K. The limiting contents of alloying elements (Ti, Hf, Nb, Ta, Cr, Mo) for the (β + γ) alloys Ni - (19-29)% Al - (22-35)%Co, are determined which allowed to avoid the formation of primary γ' -phase (decrease solidus temperature ≤ 1643 K) and hard phases of the types σ , η and δ (decrease ductility). Alloying affects the morphology of the secondary γ and γ' precipitates: globular equiaxed precipitates are formed in the alloys containing Cr, Mo, and needle precipitates are formed in alloys containing γ' -forming elements Nb, Ta and, especially, Ti and Hf. After directional solidification, (β + γ)-alloys have directed columnar special structure with a low extension of transverse grain boundaries. This microstructure allows one to increase UTS, by a factor 1,5-2 and long-term strength (time to rupture increase by a factor of 5-10 at 1173 K)

Keywords:

NiAl $\beta+\gamma$ structure alloy, mechanical properties, alloying effect, phase and microstructure idiosyncrasies

1. Introduction:

Nickel monoaluminide NiAl is considered as a basis for the design of advantageous structural cast and wrought alloys intended for service temperatures higher than those of modern nickel wrought superalloys. NiAl has a higher melting temperature ($T_m = 1913$ K compared with ≤ 1553 K for nickel alloys), low density (5.86 compared with 8-8.9 g/cm³ for nickel alloys), and high modulus of elasticity (294.2 compared with ~ 210 GPa for nickel alloys). The main factors limiting practical application of NiAl (β phase), are low-temperature brittleness, low cracking resistance at temperatures below the ductile-brittle transition temperature (T_{db}), and a relatively low strength at temperatures > 1073 K. Therefore, the main problems associated with the design of the structural alloys based on this compound are to increase high-temperature strength and simultaneously to provide a sufficient level of low-temperature ductility.

2. Principles of the design of structural β -NiAl -based alloys:

Improvement of ductility. The low-temperature brittleness of β -NiAl having an ordered bcc structure B2 (CsCl) is caused by the presence of strong covalent and ion constituents of interatomic bonds (1, 2).

The well-known method to increase the low-temperature ductility of brittle materials is their grain refinement. However, alloys with fine-grained structure have low creep resistance (3).

The opportunity to increase the NiAl ductility by the alloying of solid solution is attractive for the following reason. The world practice displays the examples of successful application of the unique effects such as the "rhenium effect" in bcc refractory metals W, Mo and Cr (4) and the "boron effect" in Ni₃Al (5) to decrease T_{db} below room temperature and to improve the deformability of these materials. Without detailed explanation of the nature of these unique effects, we only note that no alloying elements (AE) providing such outstanding improvement in ductility were found for NiAl (6,7).

The idea of solid-solution alloying for improvement of NiAl ductility is associated with the substitution of AE for either in Ni or in Al positions in their

sublattices and with the resulting changes in the character of interatomic bonds in particular, with enhancement of their metallic component or effect of AE on a degree of ordering (7,8). In practice the solid-solution alloying of the bcc B2 NiAl is accompanied by an additional low-temperature strengthening and deterioration of low-temperature ductility, as it takes place in metals, especially in bcc ones.

In this association, we considered the alternative opportunity to increase low-temperature ductility (δ) or the resistance to crack propagation (K_{Ic}) by the formation of heterophase structures, in which the ductile phase with a disordered structure forms either local inclusions, or continuous grain-boundary interlayers (9-11). Similarly to that observed another nickel aluminide, Ni₃Al (γ'), containing up to 10 - 15% of a ductile γ phase (Ni) (12).

Strengthening of β -NiAl. The effect of alloying on strength depends on the sublattice (Ni and/or Al), in which AE occupies positions and on the deviation of the initial and alloyed compositions from the stoichiometric one because the alloying affects not only the lattice distortions and the average valence of metals in intermetallics, but also the relationship between the given solid-solution composition and the limit of the AE solubility in NiAl. This is confirmed by the data (13) about solubility of AE Fe, Co, Mn, Cu, Ti, Zr, Hf, V, Nb, Ta, Cr, Mo, W in the NiAl and the data (13 - 15) showing the connection between strengthening effect of AE and positions of AE in Al/Ni sublattice of NiAl.

Taking into account the significant distinctions in the electronic structures of Ni and Al, and the difference in the physicochemical characteristics and atomic sizes of AE and Al (or Ni) to be substituted (8,14), it is reasonable to expect a nonuniform distribution of AE in the NiAl lattice, formation of the segregations clusters or the precipitations of fine particles of strengthening phases cause significant strengthening and creep resistance effects of the microalloyed NiAl single crystals (16,17). However, the ductility of these alloys is low.

Creep resistance can be substantially increased by the reduction of the area of grain boundaries (especially of transverse ones) (6,8,14).

Choice of the alloy base. With allowance for the above analysis, we intended to design a ductile high-temperature structural alloy, using the following principles: 1 - to enhance low-temperature ductility and fracture toughness by the introduction of ductile constituents into the structure of the alloy; 2 - to enhance high-temperature strength by the alloying of the β -phase solid solution and the ductile constituent with the elements strengthening both phases and decreasing the rate of diffusional processes in the alloy; and 3 -

to increase the creep resistance of the cast alloy by directional solidification decreasing the fraction of transverse boundaries.

We considered two candidate systems such as Ni-Al-Co and Ni-Al-Fe, in which NiAl is present in equilibrium with a disordered fcc Ni-solid solution and selected the Ni-Al-Co (18-20) system characterized by a sufficiently wide $\beta+\gamma$ field (8,13) and by the absence of low-temperature transformations typical of some Ni-Al-Fe alloys. In addition, Co increases the creep (10) and heat resistances of NiAl.

Below we present some data on the structure and properties of the $\beta+\gamma$ Ni-Al-Co and Ni-Al-Co-AE alloys, where AE is a transition metals of IV-VI groups of the periodic system.

3. Experimental procedure:

The $\beta+\gamma$ ternary Ni-Al-Co alloys with the compositions that in the phase diagram are present near the $\beta+\gamma/\beta+\gamma+\gamma'$ phase boundaries were selected for the study because they allowed us to determine the maximum and minimum possible contents of Al and Co, respectively, providing a stable $\beta+\gamma$ phase composition of the alloy and to estimate the optimal volume fraction of the γ phase and the effect of the γ' phase on the properties of the alloy. The alloys were prepared from 99.9% pure Al, Ni, and Co by vacuum ARC melting and by vacuum induction melting and solidification in a lined iron mold (ingots of 60 mm in diameter and 10 kg in weight for deformation and mechanical tests). The details of the melting processes are described elsewhere (19,20). Directional solidification of Ni-Al-Co-Me alloys was performed in a equipment designed in VIAM. The melt temperature was 1700°C, and the ceramic mold was cooled by emerging to molten tin (723 K) at a rate of 4 mm/min. The temperature gradient upon solidification was 200-250 K/cm. The DS samples were examined in the as-cast state (without heat treatment). Part of the ingots was extruded at 1473 K to rods of 20 mm in diameter or upset at 1473 K to a degree of deformation of 50%. The following heat treatments of cast and/or deformed samples were performed homogenization or recrystallization annealing at 1473-1573 K for 2-100 h and stabilizing annealing at 1123 K for 10-300 h (furnace or air cooling). The samples for microstructure examination were etched with the Marble reagent. A DRON-3 diffractometer with a nickel-filtered CuK_α radiation was used for X-ray diffraction. The electron-microprobe analysis was performed with a "Camebax-microbeam" device at an accelerating voltage of 20 kV and a probe current of 25 mA. The differential thermal analysis (DTA) was carried out with a Setaram device in helium

atmosphere at a heating rate of 200 K/min. Tensile tests were performed using an R-5 machine at a strain rate of 2.5 mm/min.

4. Structure and Properties of Ternary ($\beta+\gamma$) Ni-Al-Co alloys:

Microstructure and phase composition. The alloys are characterized by the presence of ($\beta+\gamma$) degenerated eutectic, which consists of layers the γ -phase around the β -phase (Fig. 1a,b,c). If the γ -phase volume fraction decreases, the hardness of the alloy increases and its density decreases (Table 1).

Table 1. Phase composition and hardness of ($\beta+\gamma$) alloys of Ni-Al-Co system in as-cast condition and after heat-treatment (1473 K, 5 h + 1373 K, 20 h).

| Alloy | Composition, at. % | | | γ -phase, vol % | | Hardness HV, MPa | | Density, kg/m ³ |
|-------|--------------------|------|------|------------------------|------------|------------------|----------|----------------------------|
| | Ni | Al | Co | cast | annealed | cast | annealed | |
| 1 | 40 | 29,6 | 30,4 | 8 \pm 3 | 0 | 3540 | 3480 | 7060 |
| 2 | 41,0 | 29,0 | 30,0 | 7 \pm 2 | — | 3500 | — | 7100 |
| 3 | 40,9 | 28,7 | 30,4 | 16 \pm 3 | 15 \pm 3 | 3510 | 3630 | 7120 |
| 4 | 42,6 | 28,7 | 28,7 | 14 \pm 3 | 14 \pm 3 | 3450 | 3870 | 7120 |
| 5 | 39,0 | 26,0 | 35,0 | 28 \pm 3 | — | 3500 | — | 7290 |
| 6 | 42,6 | 25,2 | 32,2 | 30 \pm 3 | 32 \pm 3 | 3390 | 3450 | 7340 |
| 7 | 40,9 | 24,3 | 34,8 | 33 \pm 3 | 33 \pm 3 | 3170 | 3590 | 7390 |
| 8 | 43,5 | 21,7 | 34,8 | 50 \pm 3 | 32 \pm 3 | 2990 | 3090 | 7550 |
| 9 | 36,0 | 18,0 | 46,0 | 77 \pm 3 | — | 2500 | — | 7780 |

The solidus temperature of Ni-Al-Co ($\beta+\gamma$)-alloys is about 1643-1648 K, corresponding to the temperature of the $L\leftrightarrow\beta+\gamma$ eutectic transformation in the binary Co-Al system.

The effect of deformation on structure and properties was studied in more detail for alloys 2, 5, and 9, which differ in the β/γ ratios. The alloys extruded (ratio of 1:10) at 1473 K have a heterophase hot-deformed structure (Fig. 1d,e,f), in which the β - and γ -phase fibers are elongated along the extrusion direction and consist of polyhedral grains due to recrystallization that these phases undergo upon high-temperature deformation.

High-temperature annealing (1473-1573 K) of hypoeutectic $\beta+\gamma$ alloy 2 with a small γ -phase content decreases the volume fraction of the γ phase that is

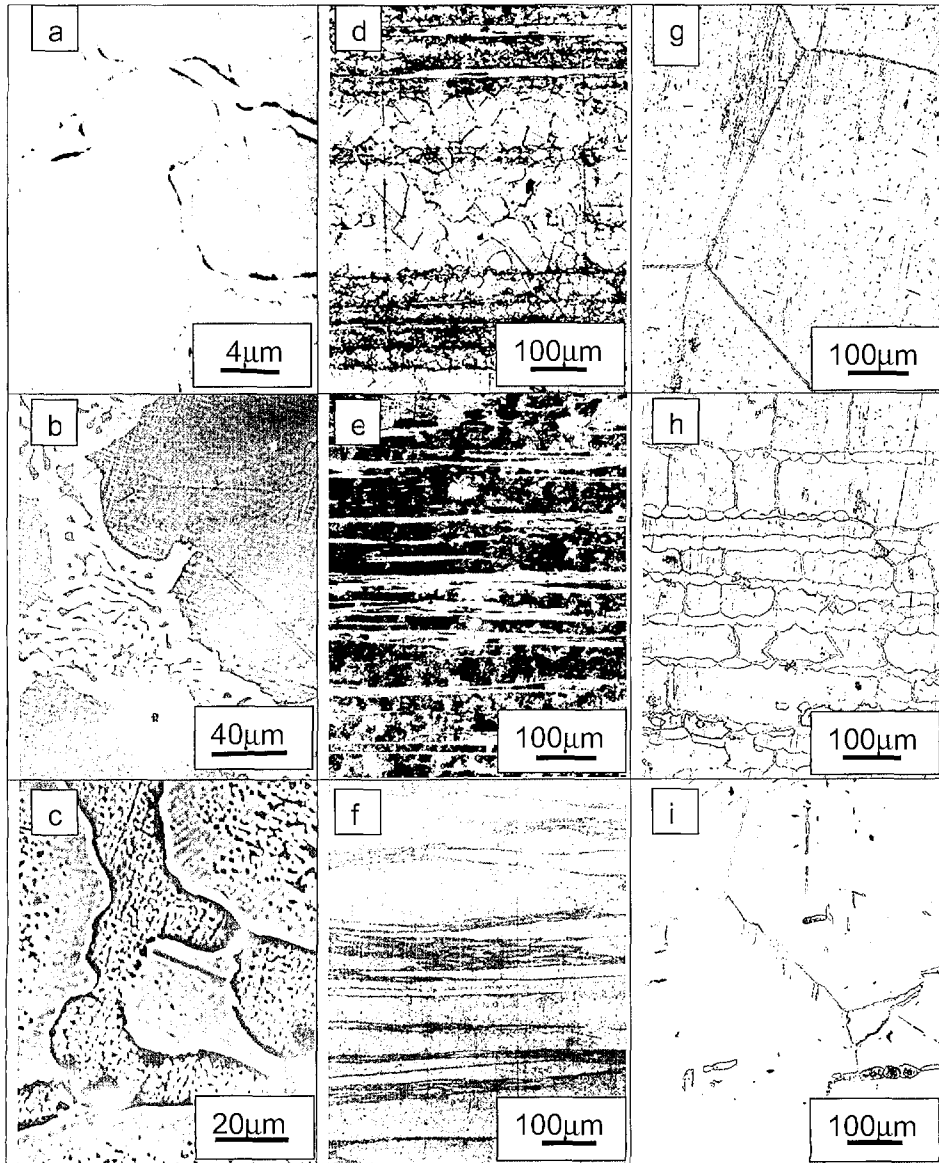


Fig. 1. Microstructure of (a), (b), (c) as-cast, (d), (e), (f) deformed and (g), (h), (i) annealed ($\beta+\gamma$) Ni-Al-Co alloys:
(a), (d), (g) alloy 2; (b), (e), (h) alloy 5; (c), (f), (i) alloy 9.

present as precipitates at the boundaries of coarse recrystallized β -phase grains (Fig. 1g). As the γ -phase volume fraction in the hypoeutectic alloy is increased to 30 vol % (alloy 5) the oriented structure is still retained, and the fibers of the γ interlayers transform to chains of recrystallized grains (Fig. 1h). These grains separate the β -phase fibers and thus hinder their broadening above 10 - 20 μm . The β -phase fibers also consist of elongated grains ($l/d \leq 3-5$), which in some cases are separated by thin transverse γ -phase interlayers. In the hypereutectic $\beta+\gamma$ alloy 9 the chains of the β -phase precipitates are present at the boundaries of coarse $\leq 500 \mu\text{m}$ recrystallized γ -phase grains (Fig. 1i). The structures of as-cast, deformed and annealed alloys 2 and 9 confirms a strong temperature dependence of the mutual solubility of the β and γ phases. The analysis of the data on the mechanical properties and density of three alloys after extrusion and heat treatment (Table 2) shows that, in the combination of properties and the most structure stability, the alloys containing ~ 30 vol % of the γ phase are advantageous as a base for the design of multicomponent wrought and, especially, cast ($\beta+\gamma$) alloys, in which the solid-solution strengthening of the β and γ phases can be realized.

Table 2. Mechanical properties of deformed Ni-Al-Co ($\beta+\gamma$) alloys after heat treatment.

| Alloy | T_{test} | UTS, MPa | YS, MPa | δ , % | ψ , % |
|-------|-------------------|--------------------|---------|--------------|------------|
| 2 | 293 | Brittle destructin | | | |
| | 1073 | — | — | — | — |
| | 1173 | 240 | 230 | 49 | 75 |
| 5 | 293 | 1060 | 490 | 31 | 31 |
| | 1073 | 240 | 230 | 33 | 54 |
| | 1173 | 100 | 90 | 114 | 40 |
| 9 | 293 | 1090 | 490 | 30 | 31 |
| | 1073 | 400 | 250 | 1 | 1 |
| | 1173 | 160 | 150 | 12 | 12 |

5. Effect of Alloying on the Phase Composition of the Ni-Al-Co Alloys:

We studied the Ni-Al-Co-AE alloys of four series on the basis of the following four ternary Ni-Co-Al alloys (at. %): (I) Ni - (28 - 29)Al - 25Co; (II) Ni - (28 - 29)Al - 30Co; (III) Ni - (25 - 27)Al - (30 - 35)Co; and (IV) Ni - (20 - 24)Al - (34 -

36)Co. The phase composition of alloys I can be $\beta+\gamma$ or $\beta+\gamma+\gamma'$, depending on temperature, and alloys II - IV belonging to the $(\beta+\gamma)$ field are characterized by gradually increasing γ -phase fraction. The typical $(\beta+\gamma)$ microstructures of the as-cast deformed ($\epsilon=50\%$), annealed (1573 K) and aged (1123 K) alloys belonging to series II and III are shown in Figs. 2.

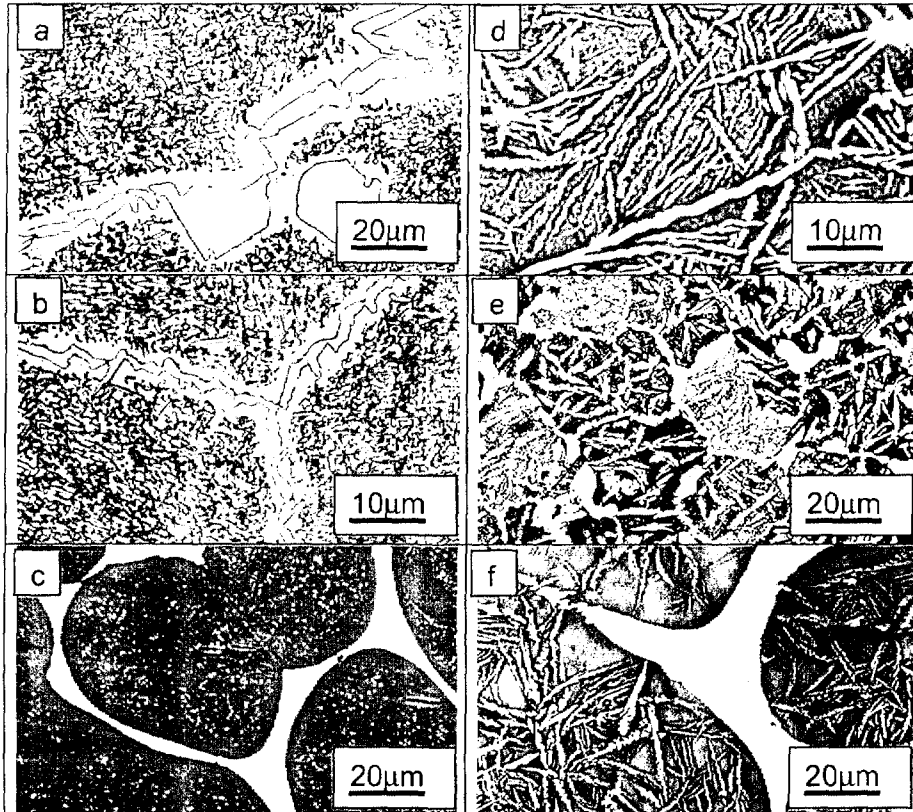


Fig. 2. Effect of the heat-treatment (e) 1573 K, 2 h and (a), (b), (c), (d), (f) 1573 K, 2 h + 1123 K, 15 h on the microstructure of deformed Ni-Al-Co-Me alloys (at. %).

(a) 39Ni-26Al-35Co, series III; (b) 41Ni-29Al-30Co, series II; (c) 38.3Ni-25.4Al-33.3Co-3.1Mo, series III; (d) 41Ni-28Al-29Co-2Ta, series III; (e), (f) 38.8Ni-26.6Al-33.5Co-2.1Ti, series III.

The effect of alloying on their phase composition analyzed on the basis of the electron-microprobe data is illustrated in Figs. 3-5.

Alloys II containing (12 - 22.3)% Cr or a small (~3%) amount of Ti and all alloys III containing (1.96 - 4.3)% Ti, (1 - 2)% V, (1.65 - 3.96)% Nb, (0.56 - 1.27)% Ta, (4.85 - 12.24)% Cr, (0.5 - 3.5)% Mo in the as-cast state have approximately similar structures (Fig. 1b), typical of ($\beta+\gamma$) alloys. The as-cast alloys with the additions of the metals of IV and V groups of periodic table (Hf, Nb, Ta and, especially, Ti) exhibit a tendency to dendritic segregation.

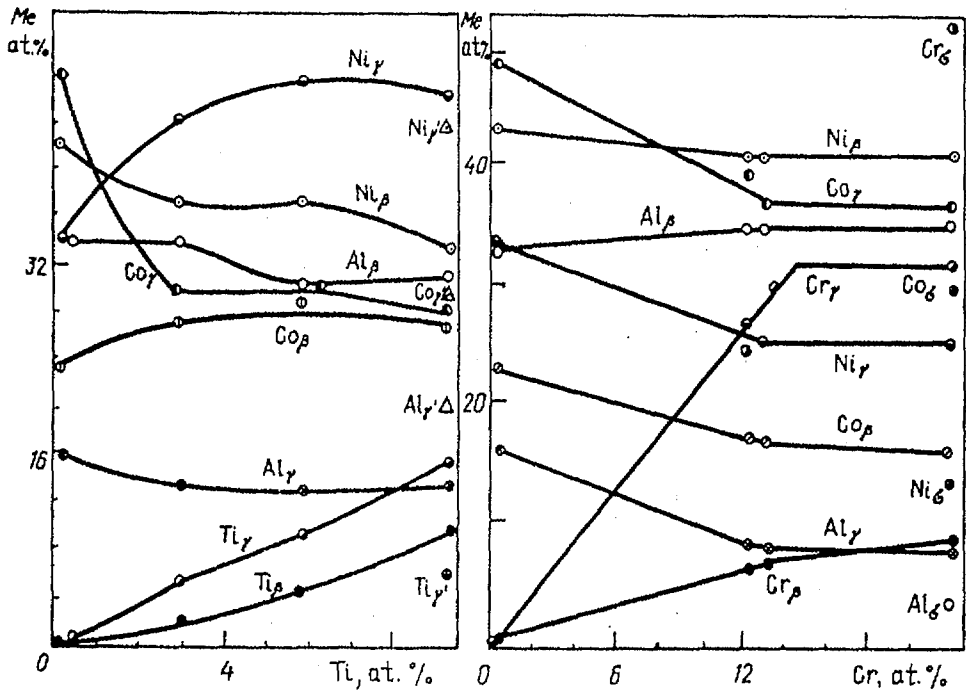


Fig. 3. Effects of titanium and chromium on the phase composition in alloys of series II.

The effect of alloying both on the morphology of secondary γ or γ' precipitates being in the β grains and on the phase composition depends on the ratio between the components in the initial alloy. As the aluminum content increases and the Co content decreases from IV to III and further II and I series, the stability of the γ -phase being in equilibrium with β -NiAl solid solution decreases in the alloys additionally containing Ti, Zr, Hf, Nb and Ta. For example, the phase composition of all alloys IV is $\beta+\gamma$ with a maximum

fraction of the γ phase. Alloys III have a smaller fraction of the γ -phase. The formation in these alloys of any phases in addition to the β and γ phases is possible only for high AE contents, whereas, in alloys II, the addition of even 1% Ta and Hf results in the disappearance of the γ -phase and in the formation of the γ' -phase. Alloys I without AE or with any AE contain the γ phase including its primary precipitates and belong either to the $\beta+\gamma$ or to the $\beta+\gamma+\gamma'$ fields. In these alloys is possible formation of TCP-phases, enriched by AE. The X-ray diffraction data on the phase composition of the alloys confirm the electron-microprobe data.

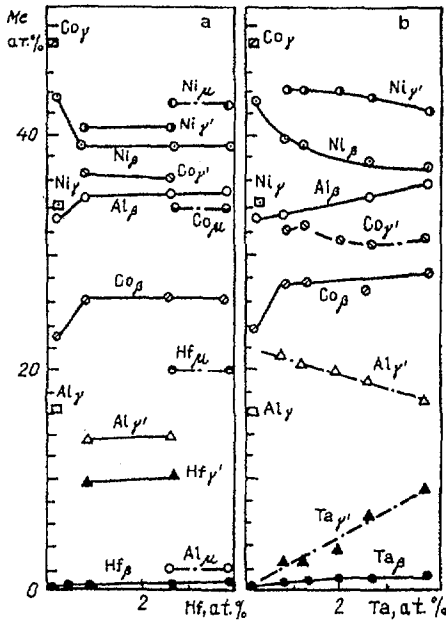


Fig. 4. Effects of (a) tantalum and (b) hafnium on the phases composition in the alloys of series II.

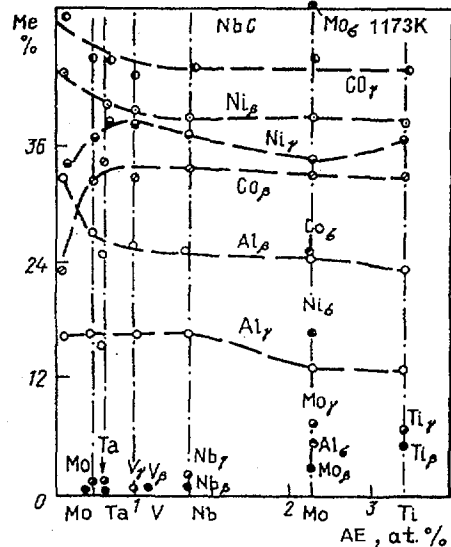


Fig. 5. Effect of alloying elements on the compositions of phases in the alloys series III.

The size and morphology of the secondary γ and/or γ' particles precipitating upon the decomposition of the supersaturated β solid solution in the alloys depend on composition AE, heat-treatment temperature and duration. Coarse elongated plate- and feather-like γ or γ' precipitates characterized by a length-to-diameter ratio l/d (or the unequixity factor UF) = $\sim 10 - 20$ and a diameter

of 0.5 - 1 μm in the cast or deformed alloys with Ti, Ta, and Nb are formed upon high-temperature (1573 K) decomposition of the β solid solution. Finer particles, precipitate upon low-temperature decomposition at 1123 K in the alloy, especially with Mo, Cr or without AE), preliminary homogenized at 1573 K are characterized by near-equiaxed shape, have $\sim 0.2\mu\text{m}$ in diameter (Fig. 2).

The solidus temperature and the solidus - liquidus temperature range ($T_S - T_L$) belong to the most important characteristics of wrought and cast alloys. The typical DTA curves for some alloys of I-IV series are presented in Fig. 6.

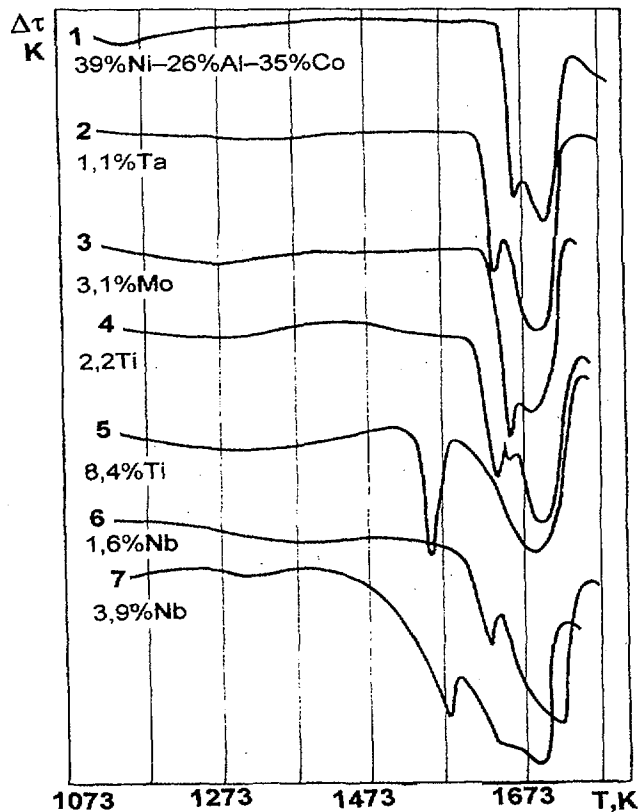


Fig. 6. Typical DTA curves for the alloys of series III with the different contents of AE (at. %).

Our main attention was concentrated on measurement of the temperature of liquid-phase formation (T_S), because it determines the upper limit of the service temperatures of the alloys. The transformations with the participation of the liquid phase can be subdivided into three groups. In the hypoeutectic alloys $\beta+(\beta+\gamma)_{eut}$ the transformations result in two "high-temperature" thermal effects (curves 1-3 in Fig. 6) that are likely to be caused by the transformations $L+\beta+\gamma \leftrightarrow \beta+\gamma$ and $L+\beta \leftrightarrow L+\beta+\gamma$. The starting melting temperature (T_S) changes from $\sim 1643 - 1653$ K for the initial ternary alloys of series I-IV to 1673-1643 K for the alloys of series III and IV with the additions of (1-2%) V, (0.56-1.27%) Ta, (1.5-3%) Mo, and (4.85-12.34%) Cr. The introduction of Ti and Nb to the alloys of series I-III (alloys with the increased Al contents) causes the appearance of an additional thermal effect at low temperatures. The correlation between the DTA data and the data on microstructure and phase composition show that the low-temperature transformation with the participation of liquid phase is caused by the formation of primary γ precipitates in the eutectic structure. The higher is the content of the γ -phase precipitates in the alloy, the lower is the starting temperature of melting. For example, this temperature varies from 1513 to 1568 K for the alloys of series I enriched with Al and depleted of Co and from 1593 to 1620 K for the γ -phase containing alloys of series III with a high content of Ta and other elements.

The effect of alloying on mechanical properties was studied for the Ni-Al-Co-AE alloys of series II and III doped with Ti, Nb Cr, Mo and Ta (Table 3).

Table 3. The effect of alloying on the mechanical properties of $(\beta+\gamma)$ alloy (wt %) 48,6Ni-10,1Al-41,3Co

| AE, wt % | UTS, MPa | YS, MPa | δ , % | ψ , % |
|-------------------------------------|--------------|--------------|--------------|--------------|
| | $293K/1193K$ | $293K/1193K$ | $293K/1193K$ | $293K/1193K$ |
| Ni-Al-Co | 1060/100 | 490/90 | 31/114 | 31/40 |
| 2,1Ti | 1240/335 | 740/290 | 12/10 | 9/12 |
| 2,4Ti 3,8Cr | 1040/270 | 620/255 | 10/8 | 10/4 |
| 2,9Ti 0,5Nb/Ta 1,0Cr 0,5Mo | 1150/330 | 650/290 | 10/8 | 6/3 |

As it is seen from the given data, the alloying of the ($\beta+\gamma$) ternary Ni-Al-Co alloy with all elements increases the characteristics of strength at room temperature and elevated temperatures. The maximal strengthening at all temperatures, especially at 1173 K, is exhibited by the alloys with the addition of ~2 wt % Ti, including those also containing the additions of 1-4 wt % Cr, 0.5 wt % Nb, Ta, and Mo.

The maximum characteristics of room-temperature ductility ($\delta = 8 - 12\%$) are exhibited by the alloys containing ~2 wt % Ti, including the multicomponent alloys. The minimum ductility ($\delta, \psi = 2 - 3\%$) is exhibited by the alloy containing 6.3 wt % Mo. This confirms the detrimental effect of brittle δ -phase precipitates. Most of ($\beta+\gamma$) alloys have a limited ability to the concentrated deformation at room temperature ($\psi = 2 - 12\%$). The similar characteristics of ductility are characteristic of heavy alloys, in which the rigid W grains are surrounded with ductile γ coats (Ni-Fe).

The presence of ductile interlayers between β grains and fine globular or extended secondary γ and γ' precipitates inside the β grains determines a high room-temperature fracture toughness (206-264 J/m²) of the alloys containing (at. %) 2,1-2,4 Ti and 3,8 Cr.

The alloys containing titanium exhibit the most stable characteristic of ductility in the entire temperature range (293-1173 K): $\delta = 8 - 12\%$ and $\psi = 10 - 12\%$ upon short-term tests and $\delta = 6 - 18\%$, $\psi = 8 - 30\%$ upon long-term tests at 1173 K. All other alloys (with Nb, Cr, and Mo) having a lower room-temperature ductility exhibit a dramatic growth of ductility with increasing temperature up to 1173 K: $\delta = 17 - 47\%$ and $\psi = 49 - 70\%$ upon short-term tests and $\delta = 30 - 50\%$ and $\psi = 30 - 50\%$ upon long-term tests. This shows that the volume fraction of the γ phase in these alloys increases and, correspondingly, their durability is deteriorated.

The durability of the alloys with a fine-grained structure containing (among other elements) titanium ($\tau=6 - 20$ h at 1173 K and $\sigma=100$ Mpa) exceeds that of the alloys containing niobium, molybdenum, and chromium. However, the mechanical properties of these alloys after deformation and recrystallization is low and can be increased by the transition to more coarse-grained cast material having either equiaxed or DS oriented structures with $l/d=10$.

This transition exhibits a distinct tendency to increasing room-temperature strength. The corresponding characteristics of ductility remain virtually unchanged or slightly increase. The durability at 1173 K of the β -NiAl+ γ + γ' alloys containing titanium increases by an order ($h\approx 600-615$ h at $\sigma=100$ MPa) and then by two orders of magnitude ($h=190$ h at $\sigma=100$ MPa) in going from the fine-grained equiaxed deformed and recrystallized structure to the coarse-

grained equiaxed cast structure and then to the coarse-grained DS structure (with a small fraction of transverse boundaries), respectively. This regularity is analogous to that exhibited by Ni superalloys. The ductility determined upon long-term tests of the alloys at 1023 and 1173 K can be characterized as moderate, ensuring the shape stability of articles upon long-term service. Especially these alloys are of interest as a ductile matrix for composition materials reinforced by single-crystal sapphire continuous fibers (21).

Conclusions:

1. New ductile NiAl-based alloys with a density of at most 7.3 g/cm^3 can be designed on the basis of $(\beta+\gamma)$ two-phase alloys (39 - 43)Ni - (20 - 25)Al - (32 - 34)Co, in which secondary γ and/or γ' -phase particles can precipitate upon heat treatment. The contents of alloying elements ($< 3\% \text{ Ti}$, $\leq 5\% \text{ Cr}$, $\leq 1-1,8\% \text{ Nb, Ta, Hf}$) are selected in such a way as to prevent the formation of the primary γ' -phase in the eutectic structure (because this decreases the solidus temperature below 1623-1643 K) and to avoid the formation of TCP phases embrittling the alloy.
2. The morphology of the secondary γ and γ' precipitates depends on the type of alloying and the precipitation temperature.
3. The alloys containing 2 - 3 wt % Ti as well as those also containing Cr (~1 wt %) and additions of Nb/Ta and Mo (0.5 wt % each) have the most interest combination of properties: high strength and stable moderate ductility in the temperature range between 293-1173 K, a high fracture toughness at 273 K. Durability at 1173 K of these alloys can be enhanced by one and two orders of magnitude by the transition to more coarse-grained cast structures either equiaxed or directionally solidified, respectively.

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