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Materials for High-Speed Transport Systems

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Abstract

Titanium aluminides are promising perspective materials which provide a unique combination of physical and mechanical properties for high-speed transport systems, like sport cars, racing cars and high-speed trains. These materials are characterized by good strength at temperatures up to 650 ^{0}C (~600 MPa), but poor room-temperature plasticity (~1-2 %). The latter property limits their commercial applications. The efforts aimed at the improvement of the plasticity of these alloys include both the design of new alloys and new methods of their production. A comparative study of the phase content, structure and the mechanical properties of the different titanium aluminides are presented. We study the influence of the crystallization conditions on the microstructure and mechanical properties of the TiAl- and Ti_3Al- based alloys prepared by the special method of pulsed volume pressing (PVP) and the Ti_2AlNb-base alloys. The factors responsible for the successful structure of TiAl-alloys with high mechanical properties were found. The cooling rate that varies in the mold with different heat capacities (0.385 cal/g K for copper and 0.12 cal/g K for steel) substantially affects the grain size, the uniformity of its distribution and the lamella thickness. In the Ti_3Al-base alloy high rates of cooling allow conserving "soft" plates of β_0 -phase. The mechanical properties and structure of the stable and metastable phases in the Ti_2AlNb-base alloys under severe plastic deformation (shear under pressure) were also studied.

Keywords: Titanium aluminides, structure, phase transformation, metastable phases, severe plastic deformation

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Introduction

The development of the high-speed transportation systems is one of the most interesting scientific directions. It is known that the materials for which produce the details for the sport cars, racing cars, high-speed trains, should work at a high temperature, pressure, and in aggressive media. Low weight is also important for these materials because it reduces fuel consumption. Titanium aluminides are the promising high temperature light weight materials which are an alternative to conventional heavy heat-resistant steels, cobalt and nickel super-alloys for automotive and aircraft engines. Titanium aluminides posses attractive properties for applications under high thermal and mechanical loads, e.g. a low specific weight, good resistance to oxidation and burn resistance at temperatures up to 1200 K, high elastic stiffness and enhanced high temperature strength, but poor room-temperature plasticity (1-2 %) (Kim, 1989; Naka et al., 1992; Appel et al., 2011). The last fact complicates their commercial application. The improvement of the plasticity of these alloys includes both the design of new alloys and new methods of production (Kazantseva et al., 2001; Kawabata et al., 1998, Kim, 2014). The optimum combination of properties can be attained by the formation of a specific structure, such as a fully lamellar two-phase TiAl/Ti_3Al structure with controlled content of γ and α_2 phases (Kim, JOM, 1989; Yamaguchi, Mat. Sci. and Technol.,1992). However, the ultimate strength, plasticity, and fracture behavior of such alloys are very sensitive to the orientation and microstructure of lamellae. It is known that a small addition of niobium or vanadium as a beta-stabilizer to the alloys based on TiAl and Ti_3Al increases room-temperature plasticity (Kawabata et al., 1998; Yao et al., 1995). The orthorhombic O (NaHg-type, Cmcm) phase of composition Ti_2AlNb, is ordered by three elements: Ti, Al, and Nb, forms in the alloys which contain ≥11at.%Nb. Ti_2AlNb-base alloys have a modified super-α_2 structure and differ in phase compositions and mechanical properties from the TiAl- base and Ti_3Al-base alloys (Bendersky et al., 1994; Boehlert et al., 1999; Kazantseva et al., 2002-2003). The improvement of the plasticity of the alloys on the base of titanium aluminides of different ages can be also achieved by special structure creation or by the use of the metastable phases. Methods of severe plastic deformation, such as shear under pressure, shock-wave loading, and equal channel angular pressing, are used successfully to obtain fine-grained structures of materials, which substantially improves their mechanical properties (Greenberg et al., 2005; Kazantseva et al., 2003).

In this report, we present a short review of the structure and mechanical properties studies of the titanium aluminides with the different chemical compositions.

Experimental

The TiAl and Ti_3Al samples were prepared by the experimental method of pulsed forging (pulsed volume pressing, PVP). The ternary alloys (Ti_2AlNb-base) were prepared by arc melting under an argon atmosphere using titanium (99.98%), high-purity aluminum (99.9%), and niobium (99.9%). The chemical compositions of the alloys under study are given in Table 1.

Table 1. *Chemical Composition (at. %) of the Alloys under Study*

No	Ti	Al	V	Nb	Mo	Cr
1	base	48	1	-		
2	base	45	1	-		
3	base	46	1.3			
4	base	34	-	1.6	0.5	0.3
5	base	22	-	26.6	-	-

For the X-ray study samples of 0.5 mm in thickness were cut from ingots and electro - polished by electrolyte with 20 ml perchloric acid and 80 ml acetic acid (T=-15°C, V=40 V). The X-ray diffraction examination was performed using a DRON-3 diffractometer with Cu Kα radiation, λ=0.15478 nm; we used the rotation of the sample. The phase composition of alloys was determined using an ASTM standard X-ray Database. The microstructure was examined using a Neophot-2 optical microscope and a JEM-200CX electron microscope. Mechanical tests of samples 3×3×4,5 mm in size were performed in the air using an INSTRON machine at a strain rate of 0,05 mm/min. Severe plastic deformations were done by the methods shear at room temperature under quasi-hydrostatic pressure using Bridgman anvils (P=10 GPa).

Results and Discussion

The alloys after the pulsed forging did not have a texture or a dendrite structure characteristic of cast TiAl and Ti_3Al (Figure 1). According to the X-ray results, the TiAl-base alloys (Alloys 1-3) contained two ordered phases TiAl (γ, $L1_0$) and Ti_3Al (α_2, $D0_{19}$), and were in a polycrystalline state. Alloy 4 consisted of three ordered phases β_0 (B2), ω ($B8_2$), and α_2 ($D0_{19}$) (Figures 2).

Figure 1. *The Cast Sample after Pulsed Forging*

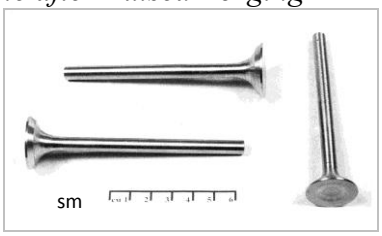
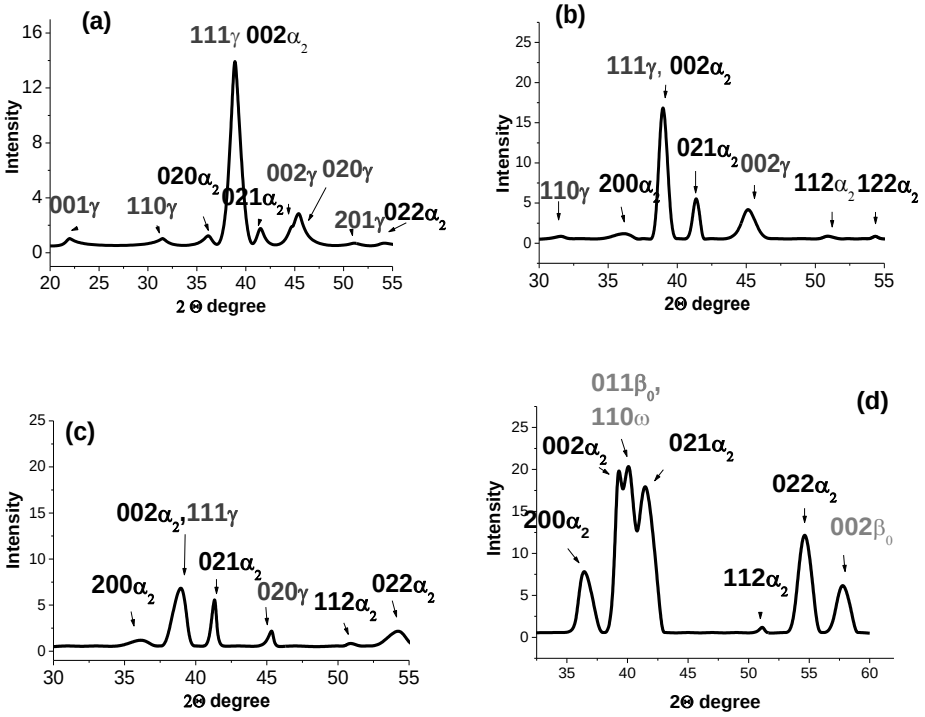
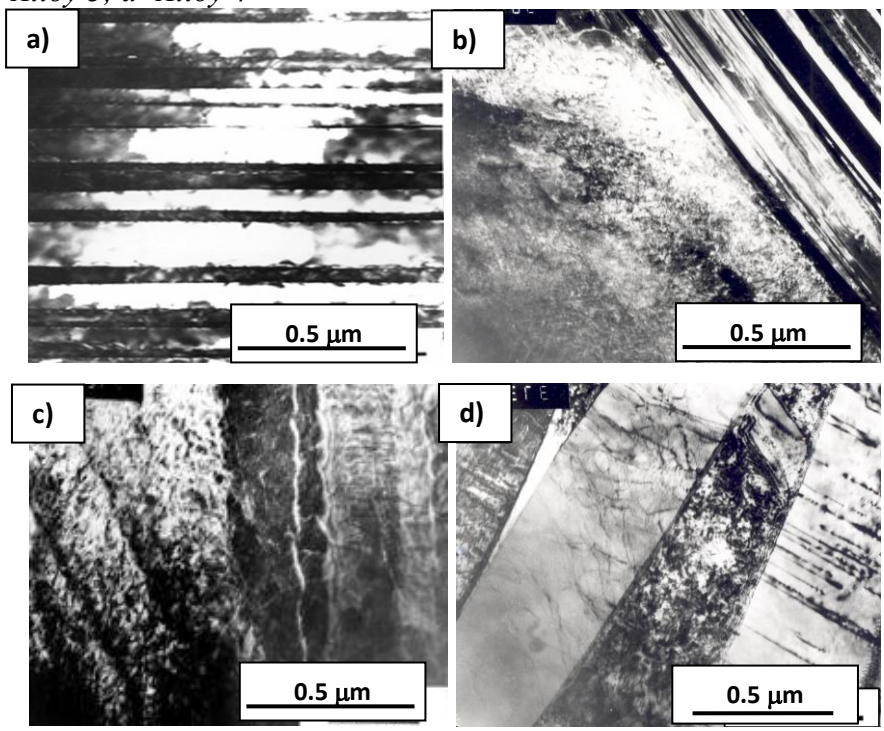


Figure 2. *X-ray Diffraction Pattern of the Samples after Pulsed Forging: a-* Alloy *1; b- Alloy 2; c- Alloy 3; d- Alloy 4*



The structure of Alloy 1 consisted on uniform equiaxial grains with an average size of about 40 μm. The lamellae inside the grains were distributed uniformly and had a small width: 0.02 to 0.06 μm for α_2 lamellae and 0.06 to 0.2 μm for γ lamellae. The structure of Alloy 2 consisted on grains with a different form. The average size of the grains was about 60 μm. The lamellar structure formed in the middle of the grains; γ and α_2 lamellae were $\approx$ 0.3 and 0.08 μm wide, respectively. The zone of the elongated grains occupied the most part of the Alloy 3. The central part of the grains had thin lamellae: for that were $\gamma \approx 0.2$ μm and for $\alpha_2 \approx 0.03$ μm in wide width. According to the Ti-Al equilibrium diagram the structure of Alloy 4 must have α_2 singe phase content. β_0-Phase forms at high temperature and transforms to low-temperature α_2- phase during usual cooling of the alloy. We suppose that the addition of the beta stabilized elements (Nb, Mo, and Cr) and the high rate of cooling of the alloy allowed us to serve β_0-phase as unstable phase at room temperature. The structure of this alloy consisted on of grains with Widmanstatten a structure inside of them (Figure 3). The grain sizes of the grains were from 50 up to 200 μm. We also found the small particles of ω-phase ($B8_2$) inside of the β_0 plates.

Figure 3. *Microstructure of the Alloys under Study, TEM: a- Alloy 1; b- Alloy 2; c- Alloy 3; d- Alloy 4*



Alloys 1-4 had high strength characteristics (Table 2). A high value of the yield stress $\sigma_{0.2} \geq 700$ MPa suggested the presence of a hard deformation mode.

Table 2. *Results of the Mechanical Tests of the Alloys under Study*

No	Compressive strength $\sigma_b^{comp.}$, МПа, $t=20^0C$	Yield stress $\sigma_{0,2}^{comp}$, МПа, $t=20^0C$	Ductility ε, %, $t=20^0C$	Compressive strength σ_b^{tens}, МПа, $t=20^0C$/ 800^0C	Ductility ε, %, $t=20^0C$/ 800^0C
1	1088	685	32.6	-	-
2	1184	1007	17	-	-
3	1030	844	18	262/ 652	0,6/1,7
4	1920	1314	10	-	-
4 (annealing)	1700	872	19	-	-

It is known that the alloys were prepared in the laboratory by zone melting and had an oriented lamellar structure (polysynthetically twinned crystals, PST) and very high mechanical properties at a grain size varying between 25 and 50 μm (Naka, 1992). The maximum strength of the PST TiAl alloys at room temperature is 1800 MPa at a plasticity of 23%, the γ interlamellar spacing in the alloy is 1.4 μm. In turn, the strength of the polycrystalline TiAl

samples prepared by complex thermomechanical treatment and having a nonoriented lamellar structure is as small as 600 MPa, and the plasticity of the alloy is about 6% (Kim, 1989; Yamaguchi, 1992). Compared to the "pure" PST TiAl alloys, the PST alloys containing vanadium exhibit a higher plasticity upon both tension and compression. The plasticity of the Ti-48,4at.%Al—0,6at.%V alloys upon compression reaches 28%. The vanadium-containing alloys are characterized by a more uniform distribution of α_2 lamellae and by a thickness of γ lamellae of 0.2-2 μm (Yao, 1995).

According to (Yamaguchi, 1992) titanium aluminides with oriented lamellar structure have bad plasticity when shear deformation proceeds across the lamellar boundaries (hard mode). When shear deformation occurs parallel to the lamellar boundaries (easy mode) plasticity increases, but strength decreases. Because of that it is difficult to create all construction detail with high mechanical properties from oriented lamella titanium aluminides. This means that the almost oriented lamellar structure of the titanium aluminides received in this study is more preferable then the oriented one.

The data obtained was compared with the strength characteristics of an orthorhombic phase Ti_2AlNb-base alloy (Alloy 5). According to the X-ray analysis, the phase composition of Alloy 5 consisted of two phases $O^{enriched}+O^{depleted}$ which formed, as we supposed, after first quenching (Figure 4). The parameters of the crystal lattices of these phases were calculated as follows: *a*1=0.6102 nm, *b*1=0.9920 nm, *c*1=0.4642 nm and *a*2=0.6096 nm, *b*2=0.9599 nm, *c*2=0.4642 nm.

This alloy had a plate structure (Figure 5) which was obtained by the special thermal treatment conditions (1200 °C-1h., water quenching + 800 °C-100h., furnace cooling +700 °C-110h., furnace cooling + 500 °C-100h. furnace cooling). We used the various heat treatments within the temperature area of O-phase existence to try to align the phase composition of the alloy, but the two O-phases structure was very stable.

Figure 4. *X-ray Diffraction Pattern of Sample 5*

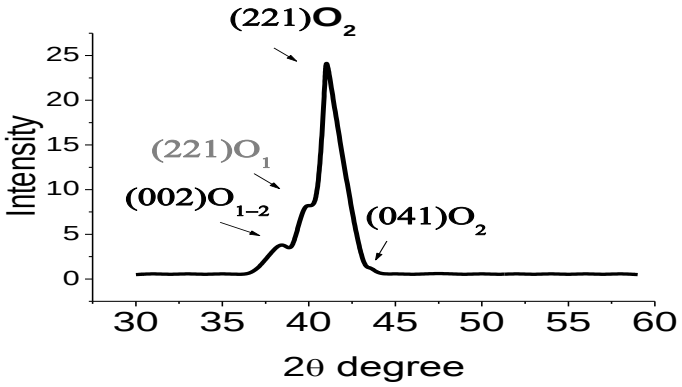
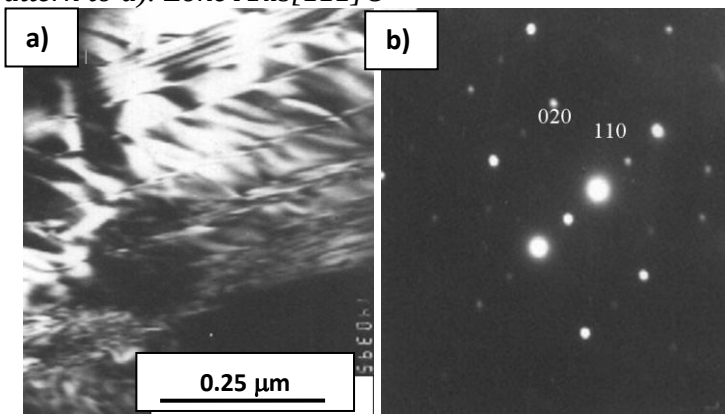


Figure 5. *Microstructure of Alloy 5, TEM: a- the Dark-field Image in (020)O; b- SAED Pattern to a). Zone Axis[111]O*

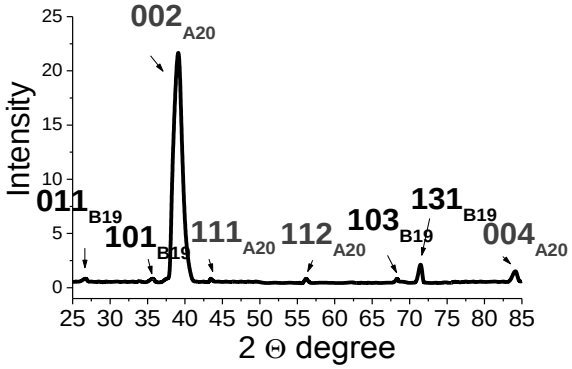


The strength characteristics of Alloy 5 with the orthorhombic phases only were as follows: $\sigma_b^{comp.} \approx 1381$ MPa, $\varepsilon \approx 19\%$. According to the literature data, the single phase (O-phase) alloys with the equaixed grains have very low mechanical properties, such as $\sigma_b^{comp.} \approx 704$ MPa, $\varepsilon \approx 1.09\%$ (Boehlert, 1999). Thus, our results show that the single phase (O-phase) alloys with the stable plate structure have higher mechanical properties in compare with the structure with the equaixed grains. Also, one can say that the mechanical properties of the Ti_2AlNb-base alloy are higher than the properties of the TiAl-base alloys (Alloys 1-3) and closer to the mechanical properties of the two phases (stable α_2 and unstable β_0) in the Ti_3Al-base alloy (Alloy 4).

Lines, which did not belong to the orthorhombic O-phase, appeared in diffraction patterns of Alloy 5 already after it was deformed to $\varepsilon = 0.9$ (hydrostatic compression). After deformation ε=4.7, the intensity of the lines belonging to the unknown phases increased. After deformation to ε=6.3, the O-phase lines were still observed in the X-ray diffraction patterns, however their intensity was substantially lower than that of the unknown-phase lines. After deformation to ε=7.6, no lines of orthorhombic phases were observed; the X-ray diffraction pattern exhibited reflections of only the unknown phases (Figure 6). Processing X-ray diffraction data by the DMPLOT program showed that severe plastic deformation leads to an order-disorder phase transformation with the formation of fine particles of two orthorhombic phases, namely, a B19 phase disordered for niobium and a completely disordered A20 phase. An analysis of the variations of the line-intensity ratio of the B19 and A20 phases allows us to conclude that, as the degree of deformation increases, the content of the B19 phase decreases, whereas the content of the A20 phase increases.

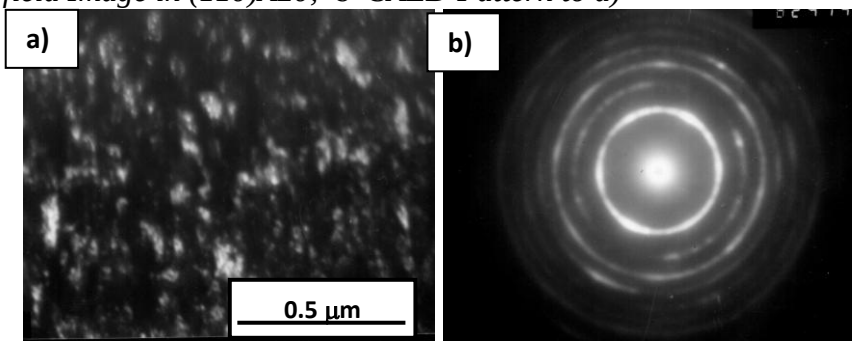
Severe deformation made for an evolution of new structure regions with new mechanical properties. Such regions may form as nano-scale fluctuation under usual deformation processes. The formation of the disordered phases effects on the strength characteristics of the alloy; microhardness of Alloy 5 was measured as follows**:** $H_{100} = 5869 \pm 40$ MPa for the initial state and $H_{100} = 4774 \pm 40$ МПа for alloy after deformation ε=7.6.

Figure 6. *X-ray Diffraction Pattern of Alloy 5 after Shear under Pressure (*ε=7.6)



TEM studies showed that with the increase of the degree of deformation the structure of the alloy becomes very fine and contains only the B19 and A20 phases. The average size of the fragments determined from dark-field images is equal to 20-30 nm (Figure 7).

Figure 7. *Microstructure of Alloy 5 after Shear under Pressure, TEM: a- the Dark-field Image in (110)A20; b- SAED Pattern to a)*



Conclusions

The study of the microstructure and mechanical properties of the titanium aluminides of the different ages leads to the following important conclusions:

1. The cooling rate, vibration, and addition pressing under crystallization were used in the experimental method. These external factors substantially affect the grain size, the uniformity of its distribution and the lamella thickness in TiAl-base alloys and should be taken into account for obtaining the alloys with the optimal structure and high mechanical properties.
2. In the Ti_3Al-base alloy high rate of cooling allows conserving "soft" plates of high-temperature β_0 phase with the metastable ω-

phase inside them. Additional aging at 900^0C-5 h. provides the increase of the plasticity by dissolution of the brittle ω-phase.

3. High temperature heat treatments allow obtaining the structure which is responsible for high strength/plasticity relation in the Ti_2AlNb-base alloys.
4. In the Ti_2AlNb-base alloys, severe deformation causes phase transformations, which occur at a fixed number. The resulting phase transformation under severe deformation in these alloys is order-disorder phase transformation with retention of crystal singony.

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