

Examining the Bimetallic Joint of Orthorhombic Titanium Aluminide and Titanium Alloy (Diffusion Welding)

Factors providing for bimetallic joint quality include absence of a continuous intermetallic layer and mutual adjustment of lattices near the contact surface

BY V. V. RYBIN, B. A. GREENBERG, O. V. ANTONOVA, L. E. KAR'KINA,
A. V. INOZEMTSEV, V. A. SEMENOV, AND A. M. PATSELOV

ABSTRACT. An excellent set of strength properties, which are inherent to orthorhombic titanium aluminides, provides a good outlook for their successful use in the development of bimetallic joints. Bimetallic joints of the orthorhombic titanium aluminide (Ti-30Al-16Nb-1Zr-1Mo) and a titanium alloy (Ti-7.7Al-1.8V) were prepared by diffusion welding. The phase composition and the microstructure of the bimetallic joint were studied using x-ray diffraction analysis, metallography, x-ray microspectrum analysis, scanning, and transmission electron microscopy. It was found that the titanium alloy mainly recovered its initial state, whereas the aluminide transformed into a disordered BCC phase upon diffusion welding. The diffusion zone had a multilayered structure. Phases forming the layers were identified. The difference between the composition of the phases and the composition of the initial alloys was due to diffusion flows of Nb and Al to the titanium alloy and Ti to the orthorhombic aluminide. The main role in the formation of the joint belonged to $\alpha_2 \rightarrow \beta$ and $O \rightarrow \beta$ transformations of intermetallic phases to a disordered β phase. This caused, first of all, the increase in the diffusion coefficients. As a result, opposite diffusional fluxes of atoms arise, which ensure adhesion of the layers near the contact surface. The mutual adjustment of the BCC lattices was a factor favoring a good-quality bimetallic joint. One more important factor was the absence of a continuous intermetallic layer near the contact surface, which could cause embrittlement. These factors were revealed earlier in our study of a bimetallic joint of the

same titanium alloy and a stainless steel, which was made by diffusion welding.

Introduction

Alloys of the Ti-Al-Nb system comprise a numerous group of alloys (Refs. 1, 2) including, on the one hand, compounds with a low concentration of Nb, which are based predominantly on the α_2 (HCP) phase, and, on the other hand, Nb-enriched compounds based on the orthorhombic O phase — Fig. 1. The O phase actually represents a weakly distorted variant of the α_2 phase. In addition to the aforementioned phases, the orthorhombic alloys may contain BCC phases depending on the composition — a disordered β phase or ordered B2 phase.

The low-temperature ductility of the O phase is higher than that of the α_2 phase and γ -TiAl (Refs. 3–6). Moreover, the O phase may be in equilibrium with the γ , α_2 , and β_0 phases. Consequently, it is relatively easy to form a wide spectrum of strength and ductility characteristics in alloys based on orthorhombic titanium aluminide by varying their chemical compositions and thermal treatment conditions.

A remarkable complex of strength properties, which are inherent to orthorhombic titanium aluminides, includes the high values of the specific strength (the strength/density ratio), good ductility at room temperature, adequate fracture

toughness, a high resistance to creep and oxidation, and good tensile properties (for certain compositions). An especially attractive feature is the possibility to make one phase as a disordered solid solution, which possesses high ductility and toughness, and, thus, can act as a buffer element in the structure of a multiphase alloy based on intermetallics. The practical application of orthorhombic titanium aluminides in real structures requires the knowledge of the phase composition, properties, and the structure of not only intermetallics, but also their fusion zones with various constructional materials.

Bimetallic joints of the orthorhombic titanium aluminide and a titanium alloy were prepared by diffusion welding (Ref. 7) and explosive welding (Ref. 8). The objective of this study was to analyze the microstructure of the joints. It was to determine which phase and structural transformations took place during diffusion welding and which factors were responsible for good-quality joints.

Experimental Procedure

The orthorhombic titanium aluminide Ti-30Al-16Nb-1Zr-1Mo and a titanium alloy (Ti-7.7Al-1.8V) were chosen as the initial materials. The alloy compositions are given in at.-%.

Ingots of the orthorhombic alloy were made by double vacuum arc remelting. Untreated ingots had an inhomogeneous coarse-grain plate-like structure with the characteristic grain size $d = 1.2$ mm. To refine grains and improve the cast structure, the alloy was treated under the following conditions. Ingots were pressed at $T = 1180^\circ\text{C}$ to 60% reduction. Then ingots were placed in an isothermal die block, which was heated to 1000°C , and were upset in three steps to 20-, 20-, and 47% reduction. As a result, forged disks 360×60 mm in size were prepared. Plates $70 \times$

KEYWORDS

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V. V. RYBIN and V. A. SEMENOV are with the Central Research Institute of Structural Materials "Prometey," St. Petersburg, Russia. B. A. GREENBERG, O. V. ANTONOVA, L. E. KAR'KINA, A. V. INOZEMTSEV, and A. M. PATSELOV are with the Institute of Metal Physics, Ural Div., Russian Academy of Sciences, Ekaterinburg, Russia.

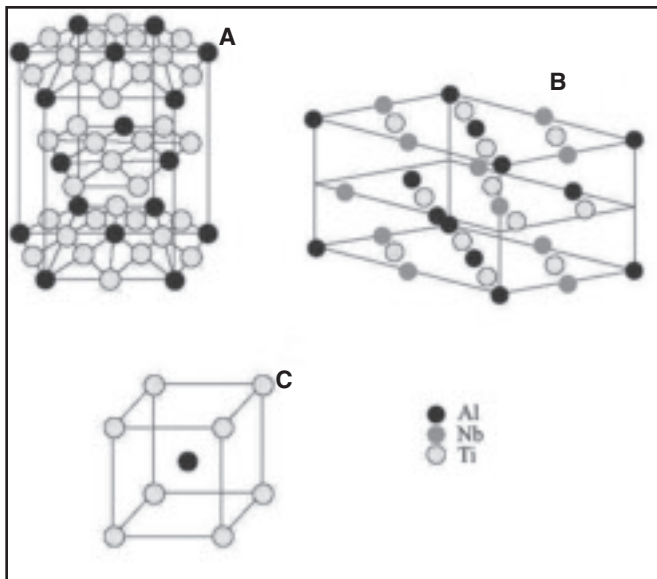


Fig. 1 — Unit cells of phases of the orthorhombic titanium aluminide. A — α_2 ; B — O; C — B2.

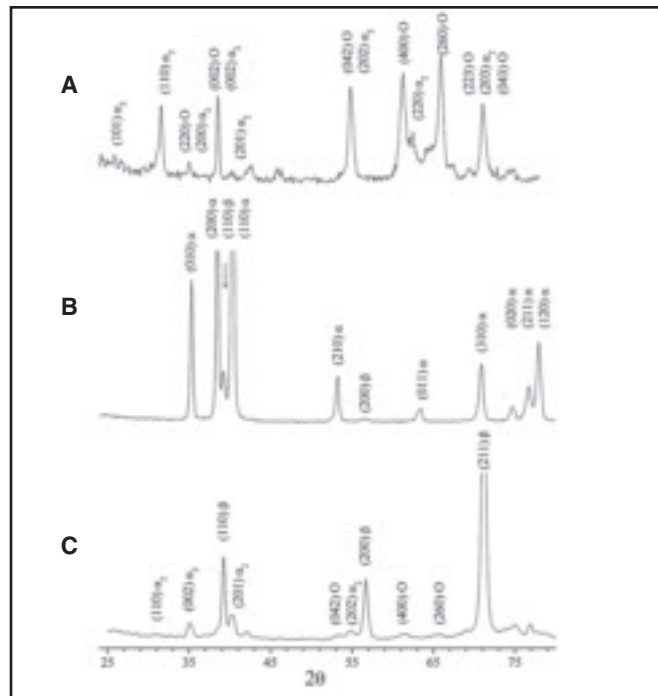


Fig. 2 — Diffraction patterns of the alloys studied. A — The initial state of the orthorhombic aluminide; B — the initial state of the titanium alloy; C — the orthorhombic aluminide after welding.

40 × 12 mm in size were cut out of the disks by the electric erosion method and rolled to a thickness of 0.5 mm. The rolled sheets were annealed under the following conditions: heating to 950°C, holding for 5 min, quenching in water, and annealing at 700°C for 3 h. Table 1 gives deformation characteristics of the alloys studied, including the yield stress $\sigma_{0.2}$, the ultimate strength σ_B , and the ductility δ .

The diffusion welding operation was performed under different loads and different temperature and time conditions. Then mechanical properties of the joint were determined. The best properties were obtained at $960 \pm 10^\circ\text{C}$ for 5 min in a vacuum of 0.133 Pa at a load of 10 MPa. This joint was studied further.

The microstructure, phases, and the redistribution of alloying elements in the joint zone were examined using optical metallography (Neophot 2, EPIQUANT with the SIAMS computing system), transmission electron microscopy (JEM 200CX), x-ray spectral microanalysis (a scanning electron microscope with an energy dispersion detector), and x-ray analysis (in monochromatized Cu-K α radiation).

Results

Initial Microstructure

Aluminide. It is seen from the diffraction pattern (Fig. 2A) that the or-

thorhombic titanium aluminide contained predominantly the α_2 phase (the D0₁₉ structure) and the orthorhombic O phase. The aluminide also included a small quantity of the BCC β phase with traces of ordering after the B2 type.

As the metallographic study showed, the structure of the orthorhombic aluminide represented regions of the globular shape against a homogeneous structural component. Regions of the globular shape had a bimodal size distribution. The coarse and fine fractions were characterized by sizes of 20 and 2 μm , respectively. The TEM examination showed that the globular regions had a layered internal structure and represented the α_2 phase at different decomposition stages, which was accompanied by precipitation of polydomains of dispersed plates of the O phase. The “background” component looked quite different. It was a duplex structure comprising relatively coarse grains of the α_2 and O phases, which filled the space between globules.

Ti-alloy. It is seen from the diffraction pattern (Fig. 2B) that the titanium alloy contained predominantly the α phase (HCP) and a small volume fraction (~8%) of the β phase (BCC), i.e., the compound represented the so-called pseudo- α -alloy (Ref. 9). The optical and transmission electron microscopic examinations revealed a lamellar structure, which is typical of the pseudo- α -alloy. Par-

allel plates of the α phase 1–3 μm thick, which were almost devoid of dislocations, formed colonies of 5 to 10 plates. Adjacent plates were separated by thin (about tenth fractions of a micrometer thick) interlayers of the β phase.

Postweld Microstructure

The x-ray spectra microanalysis of the distribution of elements in the transition zone (Fig. 3) showed that the process of diffusion welding was accompanied by intensive diffusion mixing of atoms of various chemical elements — niobium and aluminum passed from the aluminide to the titanium alloy, while titanium moved in the opposite direction. The zone of intensive diffusion mixing extended for about 10–20 μm . It is only in this zone near the contact surface (CS) that diffusing elements showed their effect on phase transformations that took place during welding.

It is seen from the diffraction pattern (Fig. 2C) that the main postweld phase of the aluminide was the β phase with traces of the α_2 and O phases. The diffraction pattern, which was obtained for the titanium alloy after welding, differed little from the initial diffraction pattern — Fig. 2A. These observations agree with results of the optical microscopy examination. The optical microphotograph (Fig. 4) shows the structure of the welded bimetal-

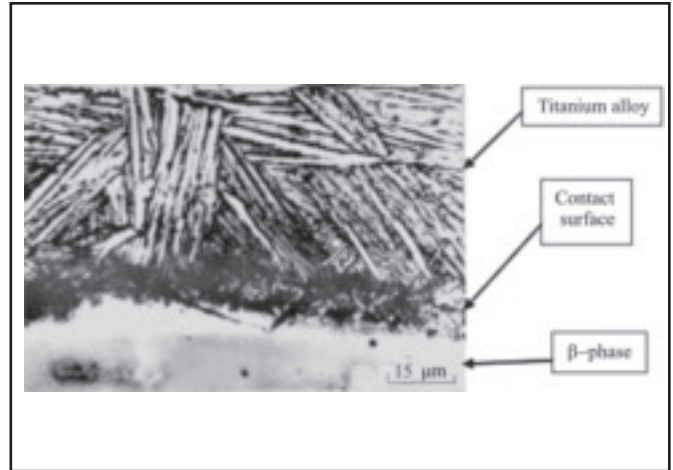
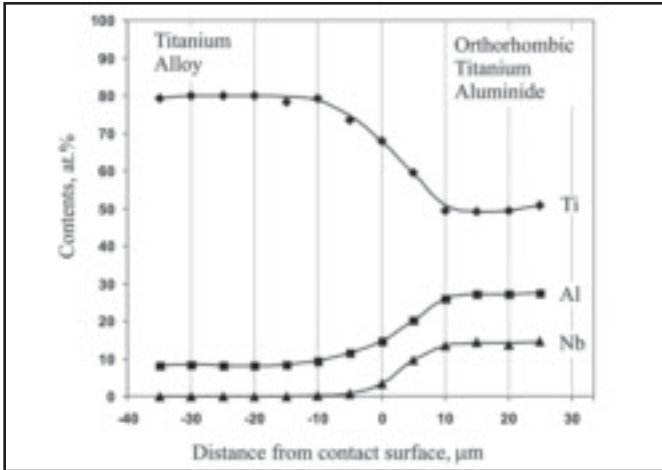


Fig. 3 — Distribution of elements in the welded joint zone as determined from the x-ray spectral microanalysis.

Fig. 4 — Optical image of the microstructure of the bimetallic joint.

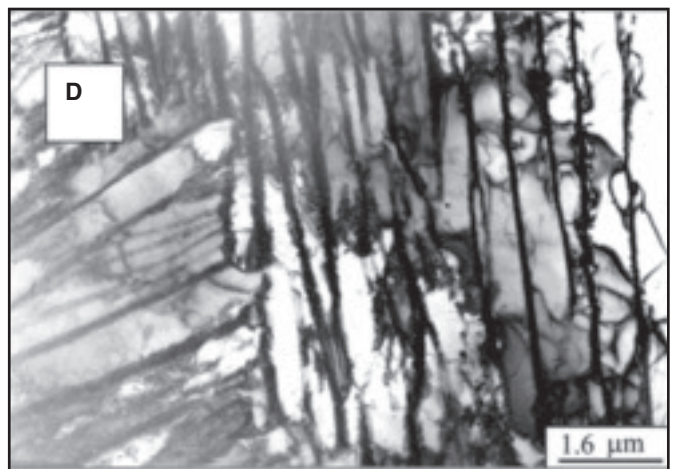
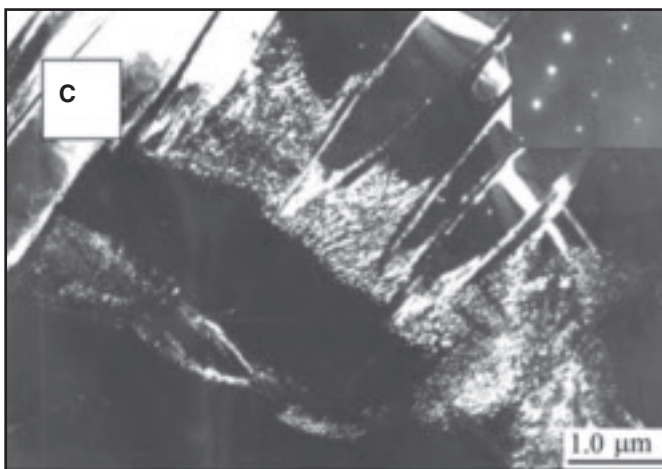
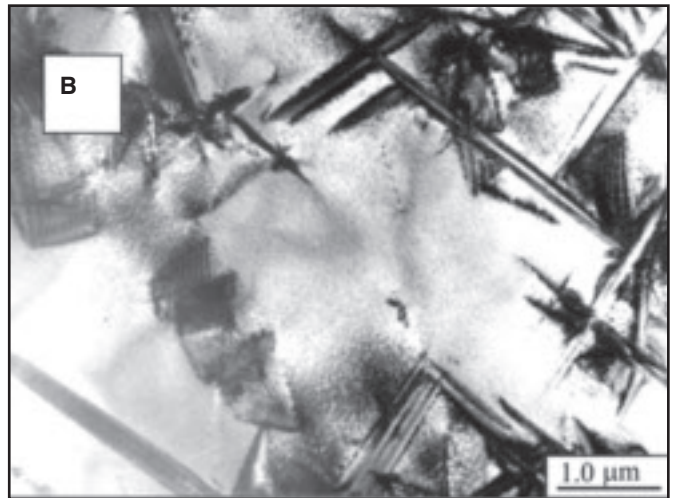
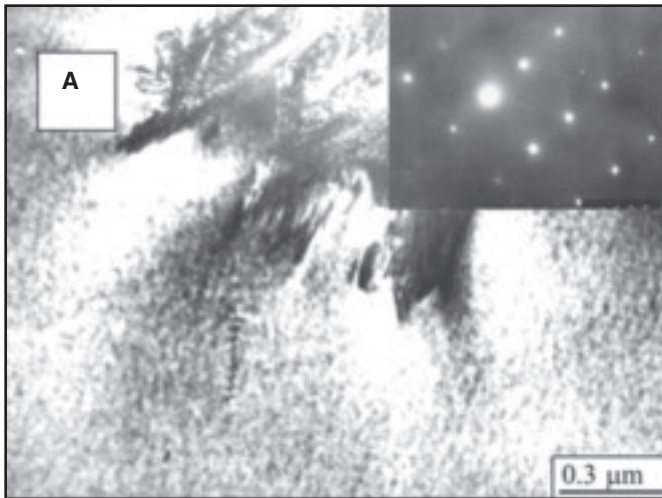


Fig. 5 — Structure of the orthorhombic aluminide far from CS (A) and near CS (B) and that of the titanium alloy near CS (C) and far from CS (D). A — Bright-field image of the β phase; B — bright-field image of the β phase with lamellar precipitates of the α_2 and O phases; C — dark-field image in the $(200)_{\alpha_2}$ reflection; D — bright-field image of plates of the α phase with interlayers of the β phase.

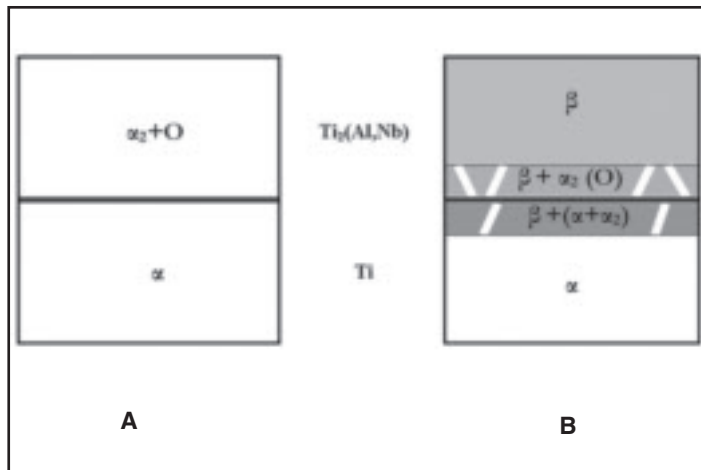


Fig. 6 — Schematic drawing of the structure of the bimetallic joint. A — Initial materials; B — after diffusion welding.

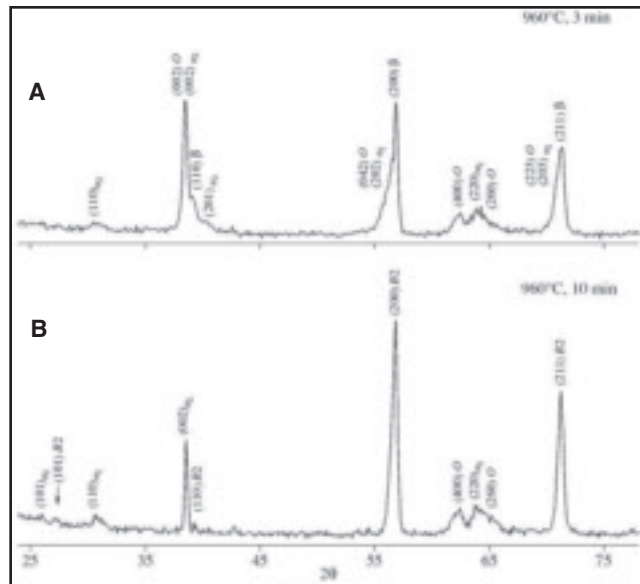


Fig. 7 — Diffraction patterns of the orthorhombic alloy after annealing at 960°C and quenching in water. A — 3 min; B — 10 min.

lic joint. The lamellar structure of the pseudo- α -alloy is seen on one side of CS. A single-phase structure of the titanium aluminide is observed on the other side.

Thus, the titanium alloy mainly recovered its initial structure (the pseudo- α -alloy), whereas the titanium aluminide turned to a disordered β phase after welding.

Welded Joint as Multilayered Sandwich (TEM)

Aluminide, far from CS. Figure 5A presents a bright-field image of the aluminide structure far from CS. The interpretation of selected-area electron diffraction patterns of this area and other areas demonstrated that these areas contained a disordered β phase.

Aluminide, near CS. Lamellar precipitates of the α_2 and O phases were seen and their volume fractions increased when CS was approached. Figure 5B presents a bright-field image of the β phase with inclusions of the α_2 and O phases. Regions of the disordered β phase were observed predominantly near CS on the side of the orthorhombic alloy.

Ti-alloy, near CS. Figure 5C presents

the microstructure of the titanium alloy in the immediate vicinity to CS. The single-phase region bounded on a colony including α plates and interlayers of the β phase. The analysis of the selected area diffraction pattern revealed that the single-phase region was filled with the β phase (BCC). Fine precipitates of the α_2 phase are seen in the α plates.

Ti-alloy, far from CS. As the distance to CS increased, the aforementioned single-phase region vanished and the β phase remained in the form of interlayers between the α phase plates, which no longer contained precipitates of the α_2 phase — Fig. 5D. At a distance from CS the structure progressively resembled the structure of the pseudo- α -alloy, which was observed before welding.

Thus, the bimetallic joint represented a multilayered “sandwich” schematically depicted in Fig. 6.

Discussion

Phase Transformations during Diffusion Welding

Aluminide. To elucidate reasons for the surprising fact that the aluminide turned

to a disordered β phase after diffusion welding, it was necessary to determine the cross section of the phase equilibrium diagram for the composition at hand. For this purpose, some test samples were annealed at temperatures from 900° to 970°C for different intervals of time and then water quenched. The obtained phase equilibrium diagram differed from the phase equilibrium diagrams (Refs. 1, 10), which contained the B2 phase of a similar composition only at temperatures exceeding 1000°–1100°C. It was shown that the alloy in the initial state was predominantly a two-phase ($\alpha_2 + O$) compound. At 900°–950°C it contained three phases ($\alpha_2 + O + B2$) and the proportion of the B2 phase increased sharply with temperature. The alloy included mainly the B2 phase and traces of the α_2 phase above 960°C. Therefore, one might expect indeed that at a temperature equal to the welding temperature, the α_2 and O phases turned, being nonequilibrium ones, to the B2 phase.

Short-time annealings were performed at 960°–970°C to see the process of the $\alpha_2 \rightarrow B2$ and $O \rightarrow B2$ phase transformations. It was found that a BCC phase was observed already after 3-min annealing, and this phase represented a disordered β phase. The corresponding diffraction pattern is given in Fig. 7A. However, lines of the B2 phase appeared already after 8- to 10-min annealing — Fig. 7B. The TEM examinations of the microstructure also confirmed the formation of a disordered BCC phase after short-time annealing and its ordering as the annealing time increased. Early stages of ordering were observed after the samples were annealed at 960°C

Table 1 — Mechanical Properties of the Alloys under Study at Room Temperature

Alloy	Direction	$\sigma_{0.2}$, MPa	σ_B , MPa	δ , %
Ti - 30Al - 16Nb - 1Zr - 1Mo	along the sheet	1150	1290	4.1
	across the sheet	1114	1291	6.6
Ti-7.7Al-1.8V		588	753	9-10

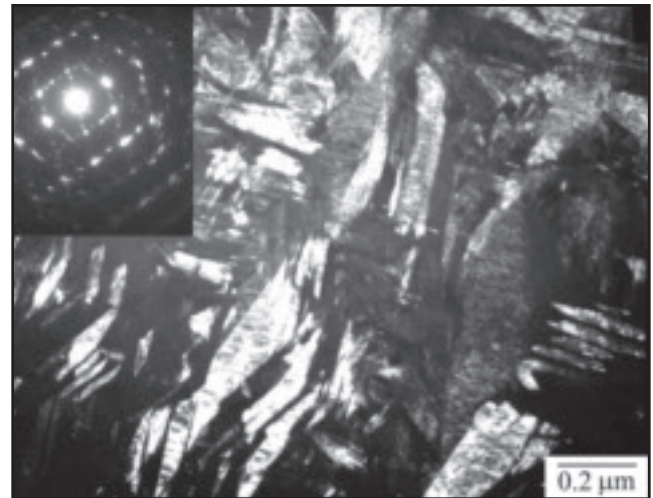
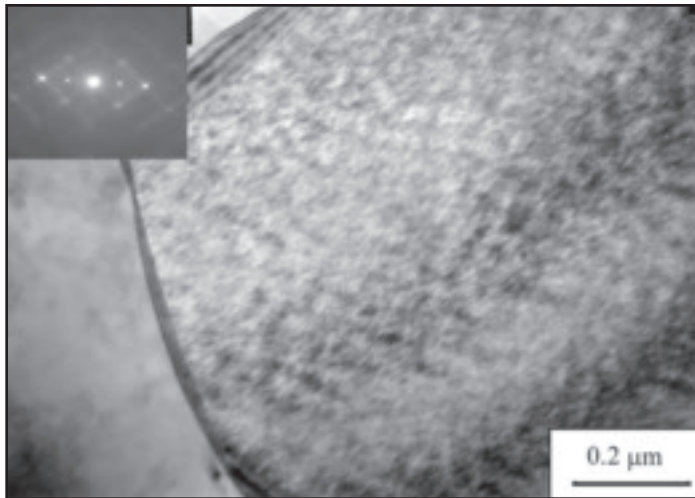


Fig. 8 — TEM image of the microstructure of the orthorhombic alloy after annealing (960°C, 8 min) and quenching in water; the B2 phase at early stages of ordering.

Fig. 9 — TEM image of the microstructure of the orthorhombic alloy after reconstruction of welding joint.

for 8 min — Fig. 8.

Thus, the $\alpha_2 \rightarrow \beta$ and $O \rightarrow \beta$ phase transformations were realized as follows: a quick transformation $\alpha_2 \rightarrow \beta$ and $O \rightarrow \beta$ to the disordered phase and subsequent ordering. The disordering process was only realized during welding — Fig. 5A. Since the aluminide composition changed near CS, namely, approached the Ti_3Al composition, lamellar inclusions of the $\alpha_2(O)$ phases could be formed in the β phase — Fig. 5B.

Ti-alloy. The diffusion welding temperature was sufficient for the $\alpha \rightarrow \beta$ transformation in the Ti alloy of the given composition. The reverse $\beta \rightarrow \alpha$ transformation could take place upon subsequent cooling — Fig. 5D. The lamellar structure, which is seen in the optical microphotograph (Fig. 4), is typical of the titanium alloy when it is cooled exactly from the β range. This also confirms the occurrence of $\alpha \rightarrow \beta \rightarrow \alpha$ transformations during welding.

The layer near CS contained not only regions having the aforementioned lamellar structure, but also regions filled with the β phase. Colonies of α plates (Fig. 5C) were enriched in aluminum. This was confirmed by the appearance of fine precipitates of the α_2 phase in the α phase plates. Moreover, the observation of the α_2 phase suggested that the Al concentration of the α phase considerably increased. Indeed, the said particles can precipitate upon cooling only if Al concentrations exceed 10 at.-%. Oppositely, isolated regions of the β phase (Fig. 5C) were enriched in Nb. In this case, Nb acted as the β stabilizer.

Factors Determining the Quality of the Welded Joint

The structures, which were observed on both sides near CS (Fig. 5B, C), had many features in common. They contained BCC β phases of the corresponding compositions and lamellar inclusions of other phases. The joint under study was formed mostly due to the $\alpha_2 \rightarrow \beta$ and $O \rightarrow \beta$ transformations of the intermetallic phases to a disordered β phase. Below is a list of the factors that create favorable conditions for a good-quality bimetallic joint.

1. The increase of the diffusion coefficients that is typical of disordering (Ref. 11). As a result, counterdiffusional fluxes of atoms arise, which ensure adhesion of the layers near CS.

2. One and the same BCC lattice on both sides of CS. This circumstance facilitated their mutual adjustment despite the fact that a continuous layer of the β phase was only observed on the side of the aluminide and separate isolated regions of this phase were present on the side of the titanium alloy. Constants of the BCC lattices in the orthorhombic aluminide and the titanium alloy were 0.3242 and 0.3249 nm, respectively. Internal stresses arising from the mismatch of the lattices can be decreased due to the formation of dislocations. This leads to a partially coherent conjugation of the phases (Ref. 11), which is facilitated by large plasticity of the disordered β phase as compared with plasticity of the initial α_2 and O phases.

3. The absence of a continuous intermetallic layer near CS. Although the BCC

phases near CS contained inclusions of the intermetallic phases, a continuous intermetallic layer, which could lead to embrittlement, did not appear.

The choice of the diffusion welding conditions proved to be successful — the heating temperature was sufficiently high so that the initial phases became nonequilibrium and started transforming to the BCC phases; the heating time was sufficiently short so that the initial phases had time only to transform to the β phase.

To restore the intermetallic phases, the welded joint was annealed under the same conditions as those used for preparation of the initial orthorhombic alloy (700°C, 3 h). The initial α_2 and O phases were actually reconstructed — Fig. 9. Continuity of the joint was not broken.

Comparison with the Structure of a Bimetallic Joint of Titanium Alloy and Stainless Steel

The structure of a bimetallic joint (diffusion welding) of a titanium alloy of the same composition as the one used in this study and a stainless steel was analyzed earlier (Ref. 12). The results are analogous to the aforementioned findings. The same lamellar structure of the pseudo- α -alloy as in Fig. 5D was observed in the titanium region far from CS. As CS was approached, β interlayers became much thicker, i.e., the concentration of the β phase increased. However, the α phase formed, as before, the basis of the alloy. However, as distinct from the aluminide joint, a continuous layer of the β phase of

the titanium alloy was present near CS. This was due to the fact that the structure of the said surface layer was determined mainly by the diffusion flow of Fe in the stainless steel joint and the diffusion flows of Nb and Al in the aluminide joint. Moreover, the diffusion zones differed considerably by their width — the diffusion zone was much wider (~40 μm) in the stainless steel joint. This was due primarily to the fact that the diffusion coefficient of Fe in $\beta\text{-Ti}$ was more than one order of magnitude larger than the diffusion coefficient of Nb (Ref. 13).

In the region of the stainless steel the initial FCC $\gamma\text{-Fe}$ phase (far from CS) transformed continuously to the BCC $\alpha\text{-Fe}$ phase (near the CS) via several layers — one layer was formed by some non-equilibrium phase, and the other layer comprised a work-hardened austenite.

It was remarkable that two BCC phases — iron-enriched $\beta\text{-titanium}$ and titanium-enriched $\alpha\text{-Fe}$ — conjugated in the immediate vicinity to CS. It was the mutual adjustment of the BCC lattices near CS that was very favorable for a good-quality joint. One more important factor was the absence of a continuous intermetallic layer. Considering relevant phase equilibrium diagrams and the obtained concentration profiles, it was reasonable to expect the formation of Ti-Fe intermetallics by diffusion mixing. Since the intermetallic was absent, then the experiment duration was insufficient for its formation.

Comparison with the Structure of a Bimetallic Joint of Orthorhombic Titanium Aluminide and Titanium (Explosion Welding)

A bimetallic joint of orthorhombic titanium aluminide and titanium was obtained (explosion welding). Orthorhombic titanium aluminide (a sheet 0.5 mm thick) comprising Ti-30Al-16Nb-1Zr-1Mo (at.-%) was used (Refs. 8, 14). The explosion welding parameters were as follows: the sheet speed of 500 m/s; the collision angle of 12–14 deg; heating temperature from surface friction and wave detonation equal to 900°C; and the pressure of 6 GPa. In the case of explosion welding, despite a strong shock-wave loading, the phase composition of the materials under study differed little from the initial composition. However, the structure changed drastically. Structural elements, which are typical of strongly deformed materials (Ref. 15), were observed on both sides of the contact surface. The concurrent presence of a cellular structure, a band structure with a high density of dislocations, and new grains pointed to different stages of fragmentation, which were due to an inhomogeneous plastic deformation over the volume of sheets. The fact that the

mixture of phases ($\text{O} + \alpha_2$) was preserved did not mean the invariability of each of the phases. It was found that phase transformations $\alpha_2 \rightarrow \text{O}$ and $\text{B2} \rightarrow \text{O}$ took place during the explosion welding. Different variants of the interrelation between recrystallization and phase transformations in the heterophase structure under shock-wave loading were considered.

Conclusions

- The structure of a bimetallic joint (diffusion welding) of the orthorhombic titanium aluminide (Ti-30Al-16Nb-1Zr-1Mo) and a titanium alloy (Ti-7.7Al-1.8V) was studied using different methods.

- The diffusion zone had a multilayered structure — the β phase (the aluminide, far from contact surface (CS)); the β phase and a lamellar mixture of ($\alpha_2 + \text{O}$) phases (the aluminide, near CS); the β phase and plates of the α phase with the α_2 phase inclusions (Ti alloy, near CS); and the β phase with interlayers of the β phase (Ti alloy, far from CS).

- It was found that the $\alpha_2 \rightarrow \text{B2}$ and $\text{O} \rightarrow \text{B2}$ phase transformations were realized as follows: a quick transformation $\alpha_2 \rightarrow \beta$ and $\text{O} \rightarrow \beta$ to the disordered phase and subsequent ordering. The disordering process was only presented during welding.

- To restore the intermetallic phases, the welded joint was annealed under the same conditions as those used for preparation of the initial orthorhombic alloy. The initial α_2 and O phases were actually reconstructed. Continuity of the joint was not broken.

- Factors that favor a good-quality bimetallic joint are as follows:

1. The increase of the diffusion coefficients that is typical of disordering;
2. One and the same BCC lattice on both sides of CS and their mutual adjustment;
3. The absence of a continuous intermetallic layer near CS.

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