BIMETALLIC JOINT OF ORTHORHOMBIC TITANIUM ALUMINIDE AND TITANIUM ALLOY: PREPARATION AND ANALYSIS OF MICROSTRUCTURE (DIFFUSION WELDING)

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ABSTRACT. An excellent set of strength properties, which are inherent in orthorhombic titanium aluminides, provides good outlooks 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 analyzed using different methods. 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 main role in the formation of the joint under study belonged to $\alpha_2 \rightarrow \beta$ and $O \rightarrow \beta$ transformations of intermetallic phases to a disordered $\beta$-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 of the 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 (1, 2) including, on the one hand, compounds with a low concentration of Nb, which are based predominantly on the $\alpha_2$ (HCP) phase, and, on the other hand, Nb-enriched compounds based on the orthorhombic O-phase. The O-phase actually represents a weakly distorted variant of the $\alpha_2$-phase. In addition to the aforementioned phases, the orthorhombic alloys may contain BCC phases depending on the composition: a disordered $\beta$-phase or an ordered B2-phase.

The low-temperature plasticity of the O-phase is higher than that of the $\alpha_2$-phase and $\gamma$-TiAl (3-5). Moreover, the $O$-phase may be in equilibrium with the $\gamma$-, $\alpha_2$- and $\beta_0$-phases. Consequently, it is relatively easy to form a wide spectrum of strength and plasticity characteristics in alloys based on orthorhombic titanium aluminium by varying their chemical composition and thermal treatment conditions.

An excellent set of strength properties, which are inherent in orthorhombic titanium aluminides, includes the high values of the specific strength (the strength/density ratio), good plasticity 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 plasticity 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. We tried to determine which phase and structural transformations take place during welding and which factors are responsible for a good quality of the joints.

MATERIALS AND METHODS

The orthorhombic titanium aluminide Ti-30Al-16Nb-1Zr-1Mo and a titanium alloy (Ti-7.7Al-1.8V) were chosen as the initial materials (6-8). 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 \( T = 1180 \) °C to 60-% reduction. Then ingots were upset forged at constant temperature 1000 °C in three steps to 20, 20 and 47% reduction to form disks with 360 mm diameter and 60 mm height. Plates \( 70 \times 40 \times 12 \) mm in size were cut out of the disks by the electric spark 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 \( \sigma_B \), and the plasticity \( \delta \).

<table>
<thead>
<tr>
<th>Alloy</th>
<th>Direction</th>
<th>( \sigma_{0.2} ), MPa</th>
<th>( \sigma_B ), MPa</th>
<th>( \delta ), %</th>
</tr>
</thead>
<tbody>
<tr>
<td>Ti–30Al–16Nb–1Zr–1Mo</td>
<td>along the sheet</td>
<td>1150</td>
<td>1290</td>
<td>4.1</td>
</tr>
<tr>
<td></td>
<td>across the sheet</td>
<td>1114</td>
<td>1291</td>
<td>6.6</td>
</tr>
<tr>
<td>Ti-7.7Al-1.8V</td>
<td></td>
<td>588</td>
<td>753</td>
<td>9-10</td>
</tr>
</tbody>
</table>

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±10 °C for 5 minutes 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\( \alpha \) radiation).

THE INITIAL STATE

The orthorhombic titanium aluminide.

It is seen from the diffraction pattern that the titanium aluminide contained predominantly the \( \alpha_2 \)-phase (the D\( \text{0}_{19} \) structure) and the orthorhombic O-phase. The aluminide also included a small quantity of the BCC \( \beta \)-phase with traces of ordering after the B2 type. The presence of this phase is verified by the peak at 20 = 56.9°. 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 μm and 2 μm respectively. The TEM examination showed that the globular regions had a layered internal structure and were identified as the \( \alpha_2 \)-phase with an incomplete decomposition, which was accompanied by precipitation of polydomains of dispersed plates of the O-phase.

**The titanium alloy.**

It is seen from the diffraction pattern that the titanium alloy contained predominantly the \( \alpha \)-phase (HCP) and a small volume fraction (~8%) of the \( \beta \)-phase (BCC), i.e. the compound represented the so-called pseudo-\( \alpha \)-alloy. The optical and transmission electron microscopic examinations revealed a lamellar structure, which is typical of the pseudo-\( \alpha \)-alloy. Parallel plates of the \( \alpha \)-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 \( \beta \)-phase.

![Graph](image.png)

Fig.1. Distribution of elements in the welded joint zone as determined from the X-ray spectral microanalysis.

**THE AFTER-WELDING STATE**

The X-ray spectrum microanalysis of the distribution of elements in the transition zone (Fig.1) 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. The main after-welding phase of the aluminide was the \( \beta \)-phase with traces of the \( \alpha_2 \)- and O-phases. The diffraction pattern, which was obtained for the titanium alloy after...
welding, differed little from the initial one. These observations agree with results of the optical microscopy examination. The optical microphotograph (Fig. 2) shows the structure of the welded bimetallic joint. The lamellar structure of the pseudo-α-alloy is seen on one side of CS. A single-phase structure of the titanium aluminate is observed on the other side.

Thus, the titanium alloy mainly recovered its initial structure (the pseudo-α-alloy), whereas the titanium aluminate transformed into a disordered β-phase after welding.

![Fig. 2. Optical image of the microstructure of the bimetallic joint](image)

**The Welded Joint As Multilayered Sandwich (TEM)**

**The aluminide, far from CS.**

Figure 3a 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.

**The aluminide, near CS.**

Lamellar precipitates of the α₂- and O-phases were seen and their volume fractions increased when CS was approached. Figure 3b presents a bright-field image of the β-phase with inclusions of the α₂- and O-phases. Regions of the disordered β-phase were observed predominantly вблизи SC on the side of the orthorhombic alloy.

**The Ti alloy, near CS.**

Figure 3c 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 α₂-phase are seen in the α-plates.
Fig. 3. 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 α₂- and O-phases; c – dark-field image in the (200)α₂ reflection; d – bright-field image of plates of the α-phase with interlayers of the β-phase

The 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 α₂-phase (Fig. 3d). At a distance from CS the structure progressively resembled the structure of the pseudo-α-alloy, which was observed before welding.

Fig. 4. Schematic drawing of the structure of the bimetallic joint:

a – initial materials; b – after diffusion welding
Thus, the bimetallic joint represented a multilayered "sandwich" (Fig.4), which included:

- the $\beta$-phase (the aluminide, far from CS);
- the $\beta$-phase and a lamellar mixture of ($\alpha_2$+O) phases (the aluminide, near CS);
- the $\beta$-phase and plates of the $\alpha$-phase with inclusions of the $\alpha_2$-phase (Ti alloy, near CS);
- the $\beta$-phase with interlayers of the $\beta$-phase (Ti alloy, far from CS).

**PHASE TRANSFORMATIONS DURING DIFFUSION WELDING**

The orthorhombic titanium aluminide.

To elucidate reasons for the surprising fact that the aluminide transformed into a disordered $\beta$-phase after diffusion welding, it was necessary to determine the cross-section of the phase equilibrium diagram for given composition. For this purpose, some test samples were annealed at temperatures from 900 to 970 °C for different interval of time and then annealed (7). The obtained phase equilibrium diagram differed from the phase equilibrium diagrams (1, 9), 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 $\alpha_2$-phase above 960 °C. Therefore, one might expect indeed that at a temperature equal to the welding temperature the $\alpha_2$- and O-phases transformed, being nonequilibrium ones, into the B2-phase.

Short-time annealings were performed at 960-970 °C to see the process of the $\alpha_2\rightarrow$B2 and O→B2 phase transformations. It was found that a BCC phase was observed already after 3-minute annealing and this phase represented a disordered $\beta$-phase. The TEM examinations of the microstructure, which were performed after different annealing periods, 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 already after the samples were annealed at 960 °C for 8 minutes.

Thus, the $\alpha_2\rightarrow$B2 and O→B2 phase transformations were realized as follows: a quick transformation to the disordered BCC $\beta$-phase and subsequent ordering. The disordering process was only complete during welding (Fig.3a). Since the aluminide composition changed near CS, namely, approached the Ti$_3$Al composition, lamellar inclusions of the $\alpha_2$(O)-phases could be formed in the $\beta$-phase.

The titanium 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. The lamellar structure, which is seen in the optical microphotograph (Fig.2), is typical of the titanium alloy when it is cooled exactly from the $\beta$-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 $\beta$-phase. Colonies of $\alpha$-plates (Fig.3c) were enriched in aluminum. This was confirmed by the appearance of fine precipitates of the $\alpha_2$-phase in the $\alpha$-phase plates. Moreover, the observation of the
\[ \alpha_2 \text{-phase suggested that the Al concentration of the } \alpha \text{-phase considerably increased. Indeed, the said particles can precipitate upon cooling only if Al concentrations exceed 10 at.\%}. \text{ Oppositely, isolated regions of the } \beta \text{-phase (Fig.3c) were enriched in Nb. In this case, Nb acted as the } \beta \text{-stabilizer.}

**FACTORS DETERMINING THE QUALITY OF THE WELDED JOINT**

The structures, which were observed on both sides near CS (Fig.3b, c), had many features in common. They contained BCC \( \beta \)-phases of the corresponding compositions and lamellar inclusions of other phases. The joint under study was formed thanks mostly to the \( \alpha_2 \rightarrow \beta \) and \( O \rightarrow B \) transformations of the intermetallic phases to a disordered \( \beta \)-phase. Let’s to list a sequences from the above that favors a good quality of the bimetallic joint.

(i) The diffusion coefficient increased that is typical of disordering (10). As a result, counter diffusional fluxes of atoms arise, which ensure adhesion of the layers near CS.

(ii) The phases on both sides of the contact surface have the same BCC lattice. This circumstance facilitated their mutual adjustment despite the fact that a continuous layer of the \( \beta \)-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 nm 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 (10), which is facilitated by large plasticity of the disordered \( \beta \)-phase as compared with plasticity of the initial \( \alpha_2 \)- and \( O \)-phases.

(iii) The absence of 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 \( \beta \)-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 \( \alpha_2 \)- and \( O \)-phases were actually reconstructed. Continuity of the joint was not broken.

**THE COMPARISON WITH THE STRUCTURE OF A BIMETALLIC JOINT OF A TITANIUM ALLOY AND A 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 (11). The results are analogous to the aforementioned findings. The same lamellar structure of the pseudo-\( \alpha \)-alloy as in Fig.3d was observed in the titanium region far from CS. As CS was approached, \( \beta \)-interlayers became much thicker, i.e. the concentration of the \( \beta \)-phase increased. However, as distinct from the aluminide joint, a continuous layer of the \( \beta \)-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 this case and the diffusion flows of Nb and Al in the previous case.

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

It was remarkable that two BCC phases – iron-enriched $\beta$-titanium and titanium-enriched $\alpha$-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, which was already mentioned in the foregoing, 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.

STRUCTURE OF 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–11Nb–1Zr–1Mo (at.% ) was used (12, 13). The explosion welding parameters were as follows: the sheet speed of 500 m/s; the collision angle of $12-14^\circ$; 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, 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 (O+$\alpha_2$) was preserved did not mean the invariability of each of the phases. It was found that phase transformations $\alpha_2\rightarrow$O and B2$\rightarrow$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.

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REFERENCES


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