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Interface characteristics in diffusion bonding of a γ-TiAl alloy to Ti–6Al–4V

XiuFeng Wang · Mo Ma · XueBin Liu · JianGuo Lin

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Abstract In the present study, diffusion bonding of a γ-TiAl alloy to a Ti–6Al–4V alloy at the different temperatures ranging from 1073 to 1173 K under an applied stress of 100 MPa for 2 h was investigated. The observation of the microstructure revealed that sound joints between the γ-TiAl Alloy and the Ti-alloy without any pores or cracks could be achieved through diffusion bonding at temperatures over 1073 K under the applied stress of 100 MPa for 2 h. The bond was composed of two zones, and its width increases with the increase of the bonding temperature. The EDS chemical composition profiles indicated that there is a diffusion flux of Al-atoms from γ-TiAl alloy towards the Ti-alloy and of Ti-atoms in the opposite direction. The microhardness of the diffusion bond was in the range of 310–450 HV, and increased monotonously from the side near the γ-TiAl alloy to the side near the Ti-alloy. In this bonding process, the diffusion flux of Ti atoms in interface is mainly controlled by grain boundary diffusion.

Introduction

γ-TiAl alloys exhibit low density, excellent high temperature strength retention, better oxidation resistance and good creep properties at high temperatures, etc. These alloys are one of the most promising materials for high temperature application in the aerospace industry [1–3]. To ensure their application and to make them more attractive for potential utilization, the development and adoption of adequate joining techniques have to be conducted simultaneously with the improvement of the properties of these alloys. Fusion welding techniques, including gas tungsten arc welding [4], laser welding [5] and electron beam welding [6], have been successfully used in the bonding of γ-TiAl alloys. However, the largest challenge for fusion welding is the high residual stress and rapid cooling rate. In comparison to fusion welding processes, diffusion bonding (DB) has the advantage that no melting of the base material occurs, which produces a drastic local change of the microstructure and related mechanical properties. Furthermore, it can avoid some typical fusion welding defects, such as hot cracking and the stress cracking susceptibility. Thus, DB is a good way to the join of γ-TiAl alloys.

The existing literature has demonstrated the possibility to join both γ-TiAl alloys and Ti-alloys either to themselves or to other special materials by diffusion bonding [7–13]. In the present study, diffusion bonding of a γ-TiAl alloy to a Ti–6Al–4V alloy at different temperatures, ranging from 1073 to 1173 K, under an applied stress of 100 MPa for 2 h is investigated, focusing on the microstructural features and composition of the interface.

X. F. Wang · M. Ma · X. B. Liu · J. G. Lin
Faculty of Material & Photoelectronic Physics, Xiangtan University, Xiangtan, Hunan 411105, P.R. China

X. F. Wang
e-mail: onexf@xtu.edu.cn

J. G. Lin (✉)
Key Laboratory of Low Dimensional Materials & Application Technology, Xiangtan University, Xiangtan, Hunan 411105, P.R. China
e-mail: lin_j_g@xtu.edu.cn; lin_j_g@163.com

Experimental Procedure

In the present study, a Ti–6Al–4V rolled bar in the annealed condition, and an as-cast titanium aluminide alloy with the nominal composition Ti–46.5Al–2.5V–2Cr–1.5Nb (in atom percent, at%) were selected as experimental materials. The TiAl alloy was hot-isostatic-pressed (HIP) at the conditions of 1523 K/172 MPa/4 h. Specimens with 8×6×4 mm were cut for DB experiments. Prior to joining, the γ-TiAl samples were heat-treated at 1353 K for 16 h, and then cooled in the furnace.

DB was conducted in a Gleeble thermal-imitation machine, and the working chamber was evacuated to 1.33×10^{-2} Pa at room temperature. The mating surfaces of the specimens were ground with SiC-paper down to grit 1200 and finally cleaned in a reagent consisting of 1 vol.%HF, 6 vol.%HNO_3 and 93 vol.%H_2O for 10 s. The mating specimens were put inside a stainless steel rectangle tube with an inner dimension of 6.1 × 4.1 mm and a wall thickness of 3 mm, as schematically illustrated in Fig. 1. DB was carried out at 1073, 1153 and 1173 K with an applied stress of 100 MPa for 2 h.

The microstructures of the base alloys and of the interfaces resulting from joining were characterized by optical microscopy (OM) and scanning electron microscopy (SEM). The chemical composition of the diffusion bond was determined quantitatively by energy dispersive spectrometry (EDS). The microhardness of the interface was measured using a HVS-1000 micro-scalerometer.

Result and Discussion

Figure 2 illustrates the microstructures of the base materials. It can be seen that the microstructure of Ti–6Al–4V alloy in the annealed condition consists of two phases: the equiaxed primary α phase, with a size of 20–30 μm, and the intergranular regions of transformed β (Fig. 2a). The γ-TiAl alloy after HIP and heat treatment at 1353 K for 16 h exhibits a near γ microstructure, with some α_2 particles or lamellar colonies at γ grain boundaries. The size of the equiaxed γ grains is in the rough of 30–50 μm (Fig. 2b).

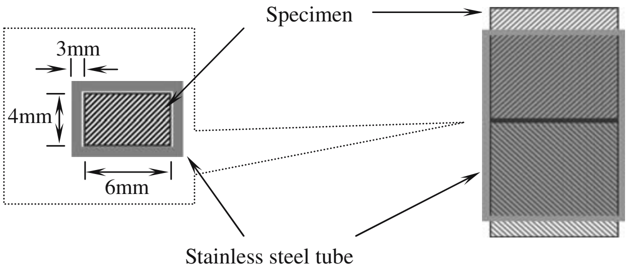

Fig. 1 Schematic diagram of the γ-TiAl/Ti–6Al–4V alloy diffusion bonding mode

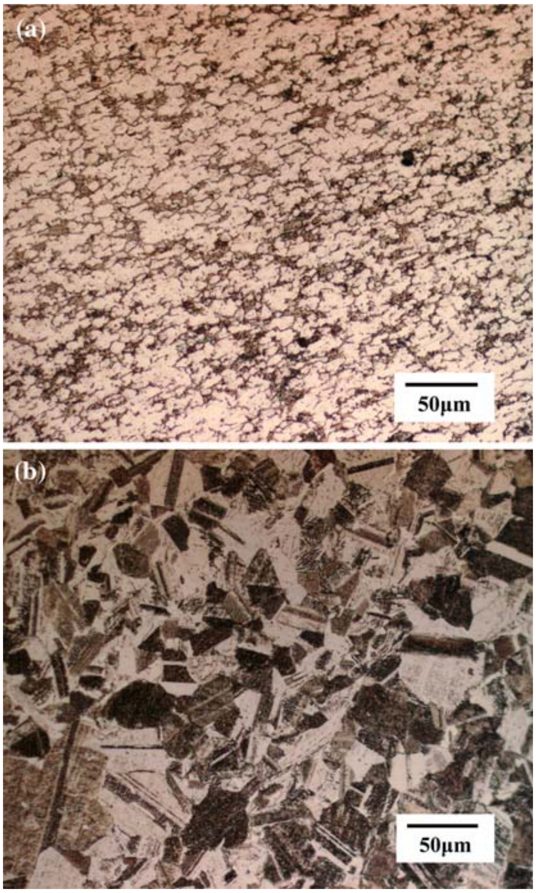

Fig. 2 The microstructure of the two base materials, (**a**) Ti–6Al–4V and (**b**) γ-TiAl alloys in annealed condition

The development of the interfacial microstructure during diffusion bonding for the different processing conditions used in this investigation is illustrated in Fig. 3. Many micro-voids can be observed at the interface after DB at 1073 K for 0.5 h. However, these micro-voids were eliminated by increasing the dwelling stage to 2 h, as it can be observed in Figs. 3b–d. Therefore, sound bonds between γ-TiAl and Ti alloys could be achieved when joining is carried out at temperatures of 1073 K and over, under an applied stress of 100 Mpa for 2 h. The bond width increased with the increase of bonding temperature, and it was measured to be 5.5, 11 and 15 μm for the samples bonded at 1073, 1153 and 1173 K, respectively. Furthermore, it is clearly visible that the joining temperature induces microstructural alterations in both

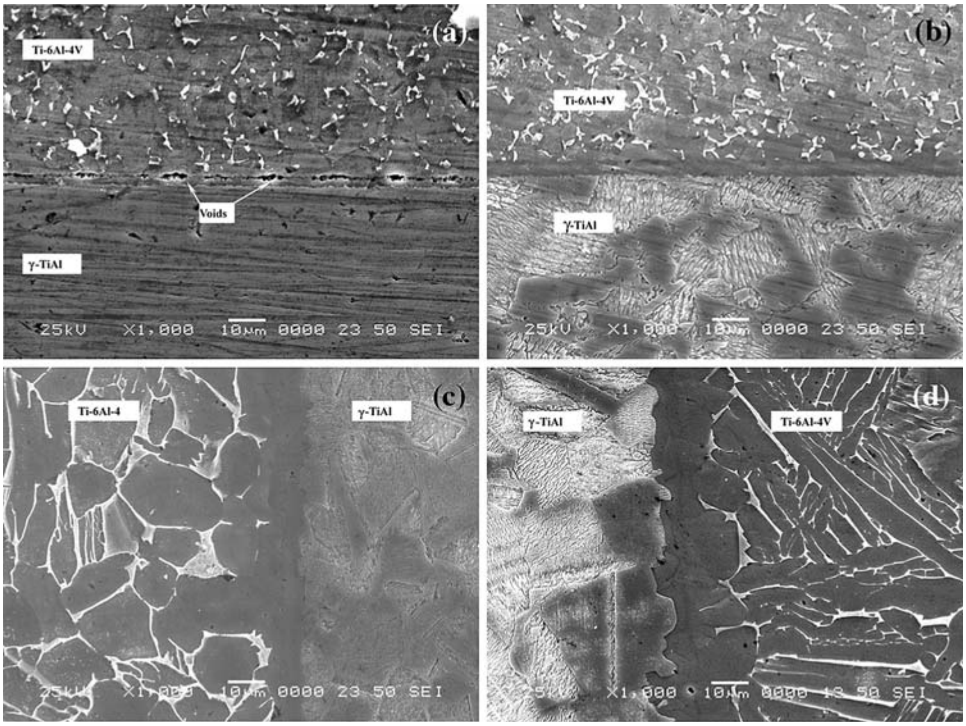

Fig. 3 The SEM microstructure of the interface for diffusion bonding under an applied stress of 100 MPa at: (**a**) 1073 K for 0.5 h, (**b**) 1073 K for 2 h, (**c**) 1153 K for 2 h, (**d**) 1173 K for 2 h

γ-TiAl and Ti–6Al–4V base alloys, i.e., the grains of these base alloys grow up when the bonding temperature was elevated.

Figure 4 is a back-scattered electron (BSE) image of the bond produced under the conditions of 1173 K/100 Mpa/2 h. It is clearly seen that the interface could be divided into two zones. Zone 1, which is located close to the γ-TiAl alloy, is generally wider than zone 2, which is near the Ti alloy. No evident boundary could be detected between the two zones, but the transition from the γ-TiAl parent alloy to zone 1 seems to be more distinct than that between zone 2 and the Ti parent alloy. The concentration profiles across the interfaces resulting from joining at the different temperatures are shown in Fig. 5. The concentration profiles indicate that there was a diffusion flux of Al atoms from γ-TiAl alloys towards Ti alloy and of Ti atoms in the opposite direction. The diffusion causes enrichment in Ti and depletion in Al in zone 1, leading to some increase in the amount of the α_2 phase. In contrast to zone 1, the net diffusion fluxes of Al and Ti atoms in zone 2 cause both its enrichment in Al and depletion in Ti, leading to the nucleation of the α_2 phase in the original α-Ti grain structure. These are similar to results reported by M. Holmquist et al. [12, 13], in which, it had been shown that zone 1 was composed of large $\gamma + \alpha_2$ grains, while zone 2 was composed of smaller $\alpha + \alpha_2$ grains.

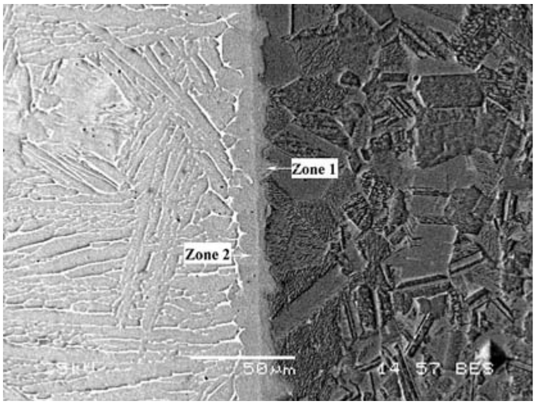

Fig. 4 The back-scattered electron (BSE) image of the interface for diffusion bonding under an applied stress of 100 MPa at 1173 K for 2 h. The micrograph shows the microstructure of the α-Ti and α_2-Ti_3 Al phases are interlaped, and appear brighter than the γ-TiAl phase

The diffusion profile of Ti in the γ-TiAl alloy can be obtained by solving the Fick's second law, assuming that two semi-infinite rods with different compositions, C_1 (i.e., Ti in γ-TiAl alloy) and C_2 (i.e., Ti in Ti-alloy), are joined together and annealed. Then we have a simple analytical relationship for the concentration profile as the following:

$$C(t,x) = \frac{(C_1 + C_2)}{2} + \frac{(C_1 - C_2)}{2}\,\mathrm{erf}\left(\frac{x}{2\sqrt{Dt}}\right) \qquad (1)$$

where $C(t, x)$ is the concentration of Ti at the interface, D is the Ti diffusion coefficient in the γ-TiAl alloy,

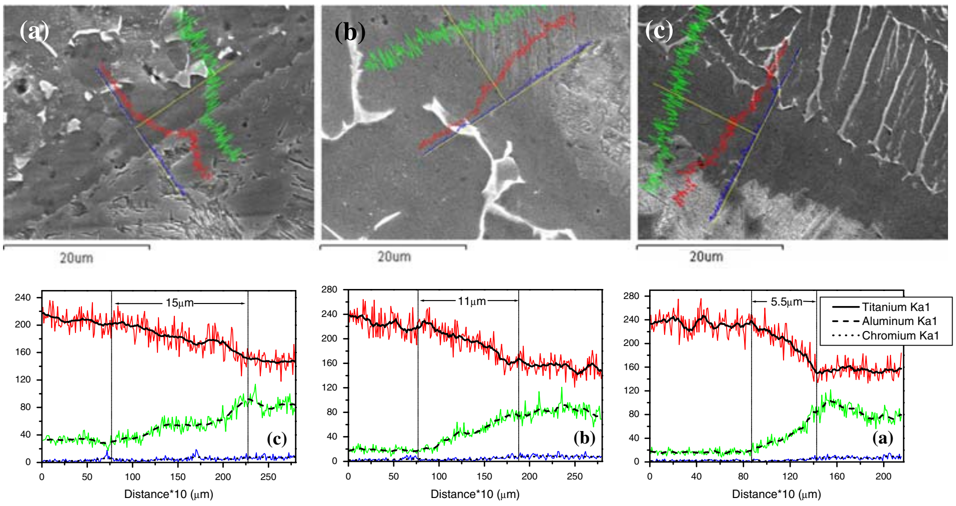

Fig. 5 Microstructure of the interface and EDS line scanning of Ti (red), Al(green) and Cr(blue) for diffusion bonding under an applied stress of 100 MPa at: (**a**) 1073 K for 2 h, (**b**) 1153 K for 2 h, (**c**) 1173 K for 2 h

which is only temperature dependent, t is bonding time and x is the distance at the interface between the γ-TiAl Alloy and the Ti alloy. Since the diffusion coefficient depends on temperature, it can be described by the following Arrhenius equation,

$$D = D_0 \exp(-Q/RT) \tag{2}$$

where D_0, Q, R and T are the frequency factor, activation energy, gas constant and temperature, respectively.

According to Eq. 1, any isoconcentration line is directly proportional to $\sqrt{Dt}$. Thus, the maximal diffusion width, $x_{\max}$, can be expressed as

$$x_{\max} = k\sqrt{Dt} \tag{3}$$

where k is constant. Therefore, by differentiating the Eq. 3, the activation energy for Ti atom diffusion across the interface can be calculated by plotting the variation in $\ln(x^2_{\max}/t)$ as a function of $R^{-1}\,T^{-1}$ for DB of γ-TiAl alloy/Ti–6Al–4V, as shown in Fig. 6. The activation energy is determined to be 200.32 kJ mol^{-1}, which is identical to the grain boundary or interfacial diffusion activation energy of 202 kJ mol^{-1} for the diffusion of Ti in γ-TiAl alloys [14]. This indicates that diffusion flux of Ti atoms in bond was primarily through the grain boundaries or the interface, which promoted the interfacial nucleation, and made the formation of fine α_2 phase grain in the interface much easier.

The Vickers microhardness across the interface resulting from joining at 1173 K/100 MPa/2 h was measured. The test results are shown in Fig. 7, where it can be seen that the microhardness of the diffusion bond is in the range of 310–450 HV, and increases monotonously from the γ-TiAl alloy towards the Ti alloy. The variation of the microhardness across the interface could be an explanation for the formation of fine α_2 phase in the bond.

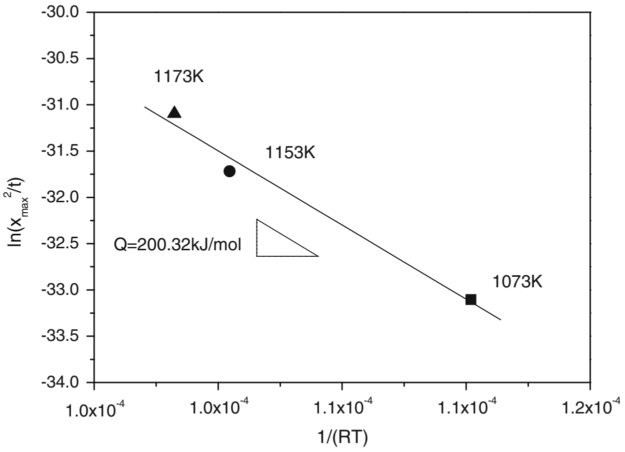

Fig. 6 Variation in $\ln(x^2_{\max}/t)$ as a function of $R^{-1}T^{-1}$ for DB of γ-TiAl alloy/Ti–6Al–4V

Conclusions

1. Sound joints between γ-TiAl Alloy and Ti alloy without any pores or cracks could be achieved through diffusion bonding performed in the temperature range of 1073–1173 K under the applied stress of 100 MPa for 2 h.
2. The interface is composed of two zones, and its width increases with the increase of the bonding temperature. The EDS chemical composition profiles show a diffusion flux of Al atoms from the γ-TiAl alloy towards the Ti alloy and of Ti atoms in the opposite direction.

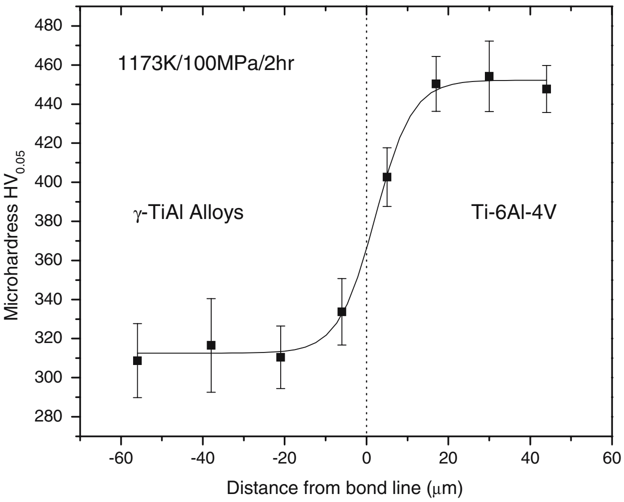

Fig. 7 Variation in microhardness $HV_{0.05}$ as a function of distance from the bond line

3. In this bonding process, the diffusion of Ti atoms in interface is mainly controlled by grain boundary diffusion.
4. The microhardness of the diffusion bond was in the range of 310–450 HV, and it increased monotonously from the size near γ-TiAl to the side near the Ti-alloy.

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