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Diffusion Welding of Al-Cu-Mg Series A2017 Alloy to Titanium†

Toshio ENJO*, Kenji IKEUCHI**, Masa-aki ANDO*** and Kunihiko HAMADA****

Abstract

Al-Cu-Mg series A2017 alloy has been diffusion-welded to titanium in the range of temperature at which the liquid and solid phases coexist in the A2017 alloy, in order to investigate effects of the coexisting liquid phase on the bond strength. The tensile strength of the joint increased remarkably with increasing the liquid phase and took a maximum value (≃240 MPa) when the volume fraction of the liquid phase was 6 ~ 7% (welding temperature ≃858 K). However, the maximum strength of the joint was considerably lower than those of the base metals of A2017 alloy and titanium. The maximum strength is considered to be determined by the strength of the intermetallic compound layer forming in the titanium adjacent to the bond interface; for the joint fractured at the intermetallic compound layer on tensile test. The intermetallic compound layer consisted of Al_3Ti and a compound containing copper.

KEY WORDS: (Aluminum Alloy) (Titanium) (Diffusion Welding) (Liquid Phase) (Intermetallic Compound)

1. Introduction

An age-hardenable Al-Cu-Mg A2017 alloy is very difficult to weld with conventional fusion welding techniques because of its high susceptibility to weld solidification cracking[1,2]. In a previous paper[3], we applied the diffusion welding to joining the A2017, and showed that joint strength comparable to the tensile strength of the base metal could be obtained in the range of welding temperature above the solidus where the liquid and solid phases coexisted.

The purpose of the present investigation is to study the effects of the coexisting liquid phase on the diffusion welding of A2017 to a dissimilar metal. The commercially pure titanium was selected as the dissimilar metal to bond with the A2017, since the titanium was a kind of light metals having a strength comparable to that of A2017 and there has been only a little amount of information about the diffusion welding of this combination.

2. Experimental Details

The chemical compositions of the base metals used are shown in **Table 1**. The base metal of A2017 was a cold-rolled plate 20 mm thick, and that of titanium a cold-drawn bar 15 mm in diameter. The specimen for the welding was a rod 15 mm in diameter and 37 mm in length prepared from the base metals by turning. The faying surface, the end of the rod, was wire-brushed (stainless wire 0.1 mm thick and 15 mm long), and subsequently degreased by washing in acetone just before the welding.

The diffusion welding was carried out in a vacuum of 2×10^{-2} Pa using an apparatus similar to that reported in a previous paper[4]; the bond zone was heated by a radiant resistance heater of molybdenum foil 0.1 mm thick, and the welding pressure to the bond interface was applied with a hydraulic press. The temperature of the bond zone was monitored with a C-A thermocouple percussion-welded to the titanium specimen about 2 mm from the bond interface, and kept at a desired temperature to an accuracy of ± 1 K.

After the welding, the joint was allowed to cool to 473 K in the vacuum and subsequently to room temperature in air. It is said that in A2017 cooled from a solu-

Table 1 Chemical compositions of base metals used (mass%).

A2017 alloy

Cu	Mg	Si	Fe	Mn	Zn	Cr	Al
4.00	0.58	0.29	0.27	0.54	0.07	0.01	Bal.

Ti

Fe	N	O	H	Ti
0.048	0.0035	0.080	0.0010	Bal.

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tion temperature the hardness obtained by ageing at room temperature is controlled by the time required to cool from 673K to 473K. In the present investigation, the cooling time from 673K to 473K after the welding was kept to be 1.14 ± 0.06ks, and the tensile test was carried out after ageing for more than 3Ms at room temperature, since the increase in the hardness was saturated at ageing time more than 3Ms.

The joint efficiency obtained was evaluated from tensile strength at room temperature. The tensile test was carried out at the strain rate of $4.7 \times 10^{-4} s^{-1}$ using an Instron-type machine. The gauge length and diameter of the specimen for tensile test was 36 mm and 7 mm, respectively. Specimens for the optical microstructure were etched with a solution of $HF : HNO_3 : H_2O = 2 : 5 : 50$. The range of temperature where the liquid phase formed in the A2017 used was determined by differential thermal analysis (DTA) which was carried out in air at heating and cooling rates of 0.05K/s with a standard specimen of nominally 99.99% aluminum.

3. Results and Discussion

3.1 Formation temperature of the liquid phase in A2017

The range of temperature where the liquid and solid phases coexisted in A2017 was determined by DTA as shown in **Figures 1 (a)** and **(b)**. In the heating process from room temperature to 850K, two endothermic reactions were observed at around 805K and above 820K. It is said that in A2017 the liquid phase starts forming at about 790K[5]. Around this temperature, a small endothermic reaction with a peak at 805K was observed as shown in Fig. 1 (a). Therefore, for the A2017 used the liquid phase presumably started forming at the lowest temperature of this endothermic reaction, 800K; accordingly, the solidus of the A2017 used lay around 800K. The amount of the liquid phase forming in this reaction, however, can be regarded as rather small, since the peak height of this reaction was small. The substantial increase in the liquid phase probably started at the temperature of 820K above which a large endothermic reaction was observed.

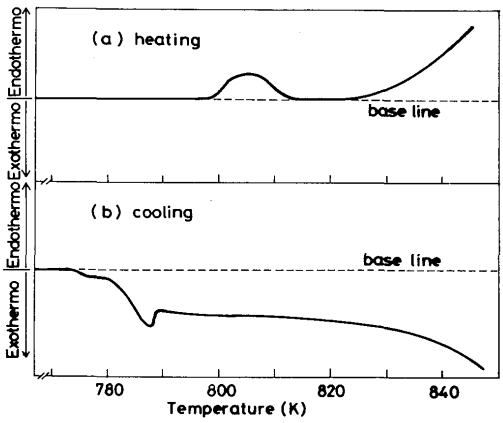

Fig. 1 Differential thermal analysis of the A2017 alloy in heating process (a) and in cooling process (b) at the heating and cooling rates of $0.05 Ks^{-1}$.

On the other hand, in the cooling process from 850K, an exothermic reaction was observed until about 773K as shown in Fig. 1 (b). This result indicated that the solidification in the cooling process continued until 773K owing to the supercooling. In order to avoid the solidification cracking in the cooling process from the welding temperature, the welding pressure was not released until 773K at which the liquid phase could be regarded as entirely annihilated; i.e., the welding pressure was exerted to compensate the tensile stress generated by the shrinkage in volume during the solidification.

In order to estimate the volume fraction of the liquid phase forming at temperatures above 820K, microstructure of the A2017 quenched from these temperatures into iced water were observed with optical microscope. As shown in **Figure 2**, a lot of dark spots and dark bands along grain boundaries were observed in the quenched A2017. These dark spots and dark bands can be regarded as the liquid phase at the quenching temperature, as described in the previous paper[3]. The volume fraction of the liquid phase was estimated from the total area of the dark spot and band. As shown in **Figure 3**, the volume fraction of the liquid phase was 3% to 11% in the range of temperature from 823K to 873K where the welding was carried out.

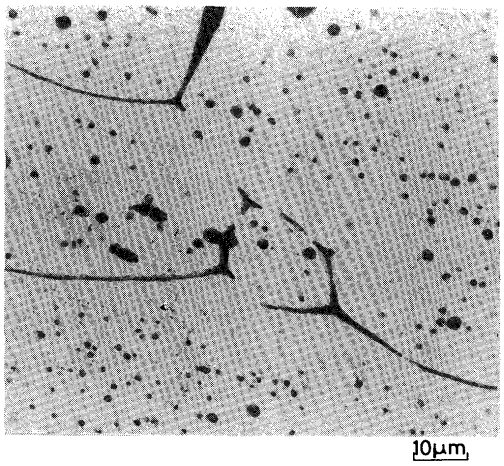

Fig. 2 Microstructure of A2017 alloy quenched from 853K into iced water.

3.2 Effects of the liquid phase on the diffusion welding of A2017 to titanium

The tensile strength of joints is shown in **Figure 4** as a function of welding temperature, together with the welding deformation. The welding deformation was estimated from the increase in the cross-sectional area of the A2017 adjacent to the bond interface. As shown in Fig. 4, the tensile strength of the joint was very low in the temperature range where the A2017 was solid state. As the liquid

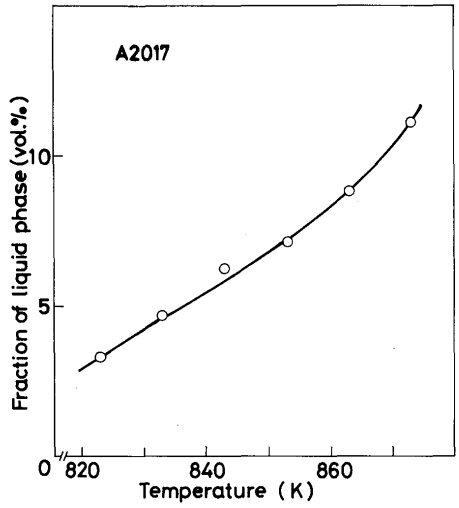

Fig. 3 Volume fraction of the liquid phase in the A2017 alloy as a function of temperature.

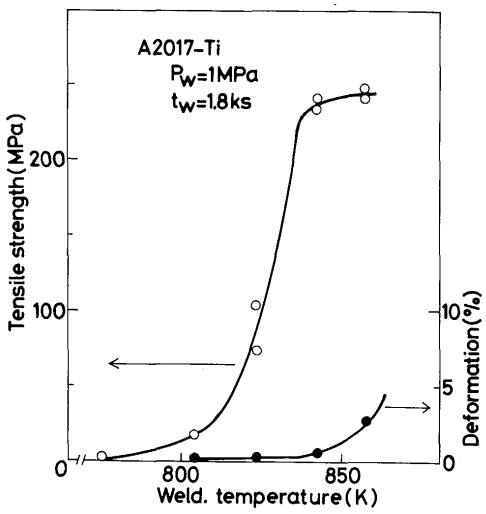

Fig. 4 Tensile strength and welding deformation of the joint of A2017 alloy to titanium as a function of temperature. Welding pressure P_W is 2MPa, and welding time t_W 1.8ks.

phase was increased, the tensile strength rised rapidly, approaching a saturated value of 230 ~ 240 MPa at temperatures of 843 to 853K. As seen in Fig. 3, the amount of the liquid phase at 843 ~ 853K was 6 ~ 8%. At welding temperatures above 863K, however, the welding deformation became too large to carry out the diffusion welding in our apparatus (minimum P_W available was 0.5 MPa). The tensile strength of joint obtained in the present investigation, though much higher than that of joint of pure aluminum to titanium[4], was considerably lower than that of the A2017 and titanium base metals; accordingly all joints obtained were fractured in the vicinity of the bond interface. In order to interpret such low joint strength, the microstructure of the joint of A2017 to titanium was investigated.

An optical microstructure of a joint welded at 853K is shown in **Figure 5**. In this joint, a marker of tungsten wire 20 μm thick was inserted between the faying surfaces

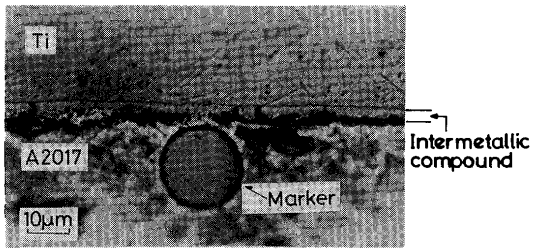

Fig. 5 Microstructure of a joint of A2017 alloy to titanium welded at 853K (P_W = 2MPa, t_W = 1.8ks). A marker of tungsten wire 20 μm in diameter is also shown.

prior to the welding in order to determine the position of the original bond interface. As shown in Fig. 5, the marker was embedded in the A2017. At the welding temperature of 853K, the flow stress of the A2017 was much lower than that of the titanium, and so the marker was quickly pressed into the A2017 by the welding pressure. Therefore, the bond interface can be regarded as contact with the titanium side of the marker. As shown in Fig. 5, an intermetallic compound layer 3 ~ 4 μm thick was observed on the titanium side of the bond interface.

A secondary electron image of this intermetallic compound layer is shown in **Figure 6**, together with distribution curves of Al, Ti and Cu analysed with EDX. The

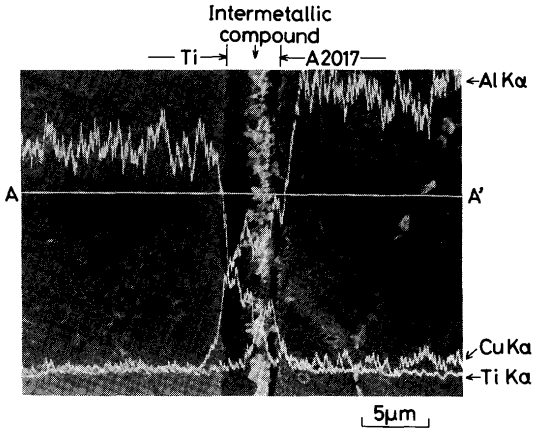

Fig. 6 Scanning electron micrograph of a joint of A2017 alloy to titanium welded at 853K (P_W = 0.5 MPa, t_W = 1.8ks). The distribution curves of aluminum, titanium and copper analysed with EDX are also shown.

slopes of the distribution curves of Al and Ti became lower in this intermetallic compound layer, suggesting that this layer consisted mainly of a compound between Al and Ti. In addition to this, many white particles were observed in the intermetallic compound layer. At these white particles the distribution curve of Cu took a maximum as seen in Fig. 6. This suggests that the white particle is a compound containing Cu. Thus an intermetallic compound containing Cu as well as that between Al and Ti seems to form in the diffusion-welded joint of A2017 to titanium.

In order to identify these intermetallic compounds, X-ray diffraction analysis using Cu-Kα was carried out. X-

ray diffraction patterns taken from fractured surfaces of A2017 and titanium sides are shown in **Figures 7(a)** and **(b)**, respectively. The diffraction lines from Al_3Ti were observed on both sides of the fractured surfaces. Therefore, the intermetallic compound layer shown in Figs. 5 and 6 can be considered to consist mainly of Al_3Ti. According to the ASTM card, no diffraction line in Fig. 7 could be assigned to a compound containing copper, though its existence was suggested by the SEM observation shown in Fig. 6. As seen in Figs. 7(a) and (b), a few diffraction lines which could not be assigned to any compound or phase composed of Al, Ti, Cu and Mg were observed at diffraction angles 2θ from 34° to 37°. It seems likely that these diffraction lines correspond to those from a compound containing copper. Thus, the diffraction lines from Al_3Ti and a compound containing

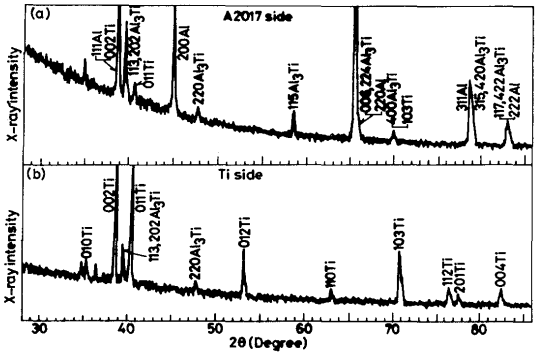

Fig. 7 X-ray diffraction patterns from fractured surfaces of a joint welded at 873 K (P_W = 1 MPa, t_W = 1.8 ks); (a) A2017 alloy side, (b) titanium side.

copper were observed on the fractured surfaces of both titanium side and A2017 side. This result suggests that the fracture of the joint occurred at least partially in the intermetallic compound layer.

In order to investigate the path of crack propagation more closely, fractured surfaces of the joint were observed with SEM. The fractured surfaces of A2017 side are shown in **Figures 8(a)** and **(b)**. For the joint welded at 823K, the fractured surface consisted of two characteristic parts. In one of them, grooves caused by the wire-brushing of the faying surface still remained, suggesting that the fracture occured at the bond interface. In the other part, the groove could not be observed. As shown in **Figure 9**, the distribution curves of Ti and Al on the fractured surface of A2017 side showed that the concentration of Ti increased at the places where the groove of wire-brushing was annihilated. Therefore, at these places the path of crack propagation can be considered to deviate from the bond interface to the titanium side.

As shown in Fig. 8(b), the place where the groove still remained could hardly be observed on the fractured surface of a joint welded at 858 K. Therefore, in this joint the fracture occurred mostly in the titanium side rather than

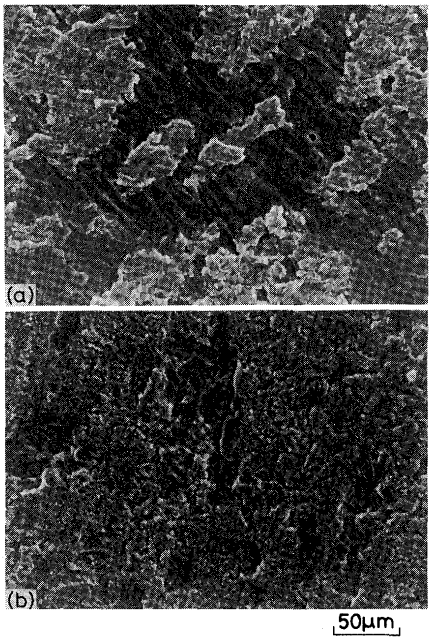

Fig. 8 Fractured surfaces of a joint after tensile test (A2017 alloy side). Welding temperature is 823 K for (a) and 858 K for (b) (P_W = 2 MPa, t_W = 1.8 ks).

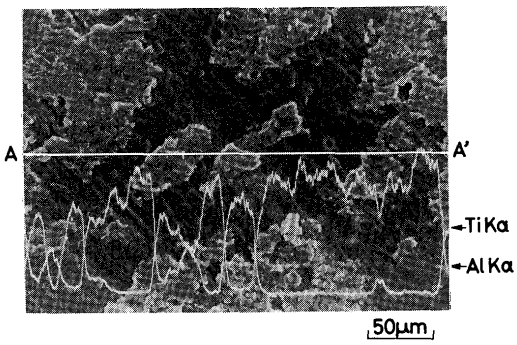

Fig. 9 Fractured surface of a joint (A2017 alloy side). Distribution curves of titanium and aluminum analysed along the line A-A' are also shown.

at the bond interface.

As shown in Figs. 8(a) and (b), the fracture morphology was brittle even at places where the fracture did not occur at the bond interface. This fact suggests the existence of a brittle phase on the titanium side of the bond interface. As seen in Fig. 5, there was no brittle phase other than the intermetallic compound layer on the titanium side of the bond interface. Thus, the observation of the fractured surface also suggests that the fracture occurs in the intermetallic compound layer.

In contrast to this, for the diffusion welding of commercially pure aluminum A1050 to titanium, joint strength more than that of the aluminum base metal (60 ~ 70 MPa) was obtained, though an intermetallic compound layer consisting of Al_3Ti was formed. This fact implys that the intermetallic compound layer of Al_3Ti did not reduce the joint strength of A1050 to titanium. Thus the effect of the intermetallic compound layer on the

joint strength of A2017 to titanium was quite different from that of A1050 to titanium. This difference can be accounted for by the comparison of the strength of the intermetallic compound layer with those of A2017 and A1050 as follows. The strength of the intermetallic compound layer was lower than that of A2017, though higher than that of A1050, and so fracture occurred in the intermetallic compound layer before the fracture in A2017. Thus, the effect of the intermetallic compound layer on the joint strength is thought to depend on the strength of the base metal. Another interpretation of the difference in the effect of the intermetallic compound layer on the joint strength is also possible considering that the intermetallic compound layer in the joint with A2017 includes a compound containing copper; i.e., the formation of the compound containing copper probably lowered the strength of the intermetallic compound layer.

4. Conclusion

The diffusion welding of A2017 to titanium has been carried out at welding temperatures above the solidus of A2017, in order to investigate the effects of the liquid phase in A2017 on the joint strength. Results obtained are summarized as follows:

(1) The solidus of the A2017 used was estimated to be ~ 800K by differential thermal analysis. The volume fraction of the liquid phase was 3% to 11% in the temperature range from 823K to 873K.

(2) The joint strength increased remarkably as the fraction of the liquid phase was increased. The maximum tensile strength of joint 240MPa was obtained at the welding temperature of 858K, that is, when the volume fraction of the liquid phase was 6 ~ 7%. At welding temperatures above 860K, the degree of the deformation of A2017 specimen during the welding became too large to perform the welding.

(3) An intermetallic compound layer about 4 μm thick was formed in the titanium adjacent to the bond interface. The intermetallic compound layer consisted of Al_3Ti and a compound containing copper. All joints obtained were fractured at the bond interface or in the intermetallic compound layer. Therefore, the intermetallic compound layer is thought to suppress the increase in the joint strength of A2017 to titanium.

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