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Influence of silicon in aluminium on the mechanical properties of titanium/aluminium friction joints

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The influence of post-weld heat-treatment and of residual silicon in aluminium on the mechanical properties of dissimilar friction joints between titanium and aluminium was investigated. Although joint tensile strength and bend test properties were drastically reduced following post-weld heat treatment, the responses of Ti/h.p. Al and Ti/c.p. Al joints were quite different. The tensile strength and bend test properties of Ti/h.p. Al joints were markedly decreased by heat-treatments involving shorter holding times at lower temperatures.

Joint failure in post-weld heat-treated joints was associated with Al_3Ti formation at the bondline region. The growth rate of the Al_3Ti intermetallic layer at the joint interface was much faster in post weld heat-treated Ti/h.p. joints. More than 20 at%Si segregated in the region between the titanium substrate and the Al_3Ti intermetallic phase in heat-treated Ti/c.p. Al joints. It is suggested that silicon segregation retards Al_3Ti formation by acting as a barrier to titanium and aluminium diffusion at the joint interface.

1. Introduction

Because of their superior mechanical and metallurgical properties, titanium and its alloys may be used in a wide range of industrial applications [1]. However, titanium is costly and this limits its application. In addition, fusion welds between titanium and metals such as aluminium or stainless steel exhibit inferior mechanical properties due to the formation of brittle intermetallic phases in completed weld deposits. Intermetallic formation has also been confirmed in dissimilar joints produced using solid-state bonding techniques –in diffusion-welded Ti/Al joints [2] and in friction welded Ti/AISI 304L stainless steel [3] and Ti/Al joints [4]. The present paper is part of a general study that is evaluating intermetallic phase formation during dissimilar friction welding of titanium and aluminium. In previous work [4], the effects of post-weld heat treatment on the mechanical properties of dissimilar friction welds between titanium and commercially pure aluminium were investigated. Al_3Ti formation was observed at the joint interface and silicon concentrated in the region between the Al_3Ti intermetallic phase and the titanium substrate. It is well-documented that impurity elements markedly influence interdiffusion at the joint interface [5] and consequently it is likely that silicon content in the commercially pure aluminium substrate may have had an important effect on joint mechanical properties, particularly on the mechanical properties following post-weld heat treatment. With this in mind, the present paper examines friction joining of titanium and low silicon content aluminium and compares the results with those produced when joining commercially pure aluminium and titanium substrates.

2. Experimental procedure

The chemical analyses of the 12-mm diameter titanium and aluminium substrates are shown in Table I. Two aluminium substrates were employed in this study – one containing 6×10^{-4} wt % Si and the other containing 0.12 wt % Si. For the purposes of the present paper, the high and low silicon content aluminium substrates are referred to as c.p. Al and h.p. Al.

A direct-drive friction welding device was employed throughout and all joints were produced using the following welding parameter settings: friction pressure 50 MPa; friction time 2 s; upset pressure 100 MPa and upsetting time 6 s. The contacting surfaces of the titanium and aluminium substrates were polished using emery paper in order to minimize the influence of surface topography on final joint properties and all

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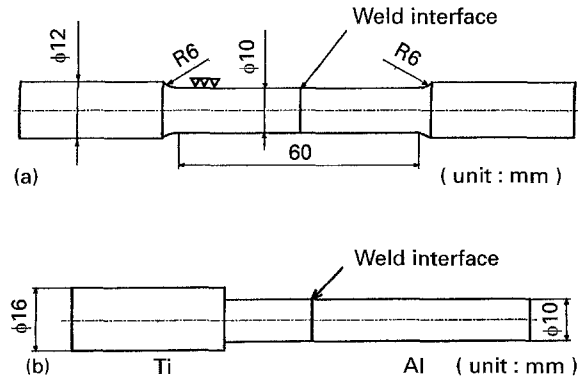

Figure 1 Dimensions of tensile (a) and bend (b) test specimens.

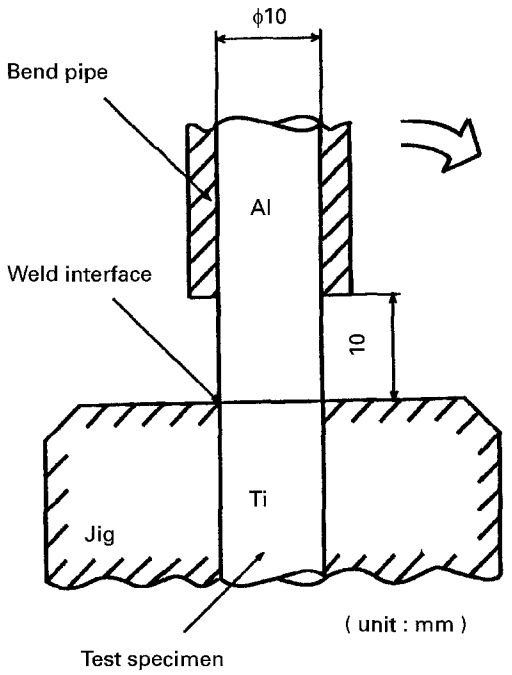

Figure 2 Bend testing method.

test specimens were degreased using acetone prior to joining.

Completed joints were mechanically-tested in the as-welded condition and following post weld heat-treatment *in vacuo* at 823 K and 873 K. During post weld heat-treatment, the holding times at temperature ranged from 0.1 h to 100 h. The heating rate during post-weld heat-treatment was 2 K s^{-1} and all test samples were immediately air-cooled to room temperature following known holding times. Joint mechanical properties were evaluated using tensile testing and bend testing. The tensile specimens were 10 mm diameter × 60 mm long (Fig. 1). Because of the widely different strengths of the titanium and aluminium substrates, three-point or four-point bending could not be applied satisfactorily. Consequently, 10 mm diameter test joints were evaluated using 90° bend testing (see Fig. 2). The maximum angle attained during bend testing was measured and the incidence of cracking at the joint interface was investigated in each case.

The joint interface region was examined using a combination of optical and scanning electron

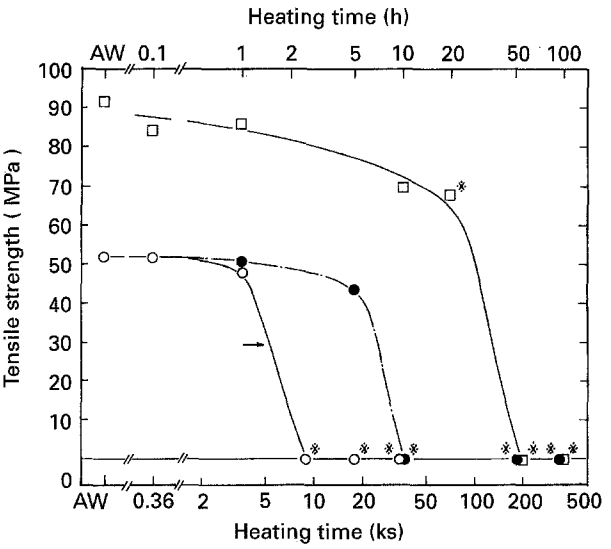

Figure 3 Effect of holding time at 823 and 873 K on the tensile strength of Ti/h.p. Al and Ti/c.p. Al joints. □ 873 K, Ti/c.p. Al; ○ 873 K, Ti/h.p. Al; ● 823 K, Ti/h.p. Al; * fractured at interface.

microscopy (SEM), transmission electron microscopy (TEM) and X-ray diffraction (XRD) analysis. TEM was carried out using a 200 kV Jeol JEM-2010 device equipped with a Noran chemical analysis attachment (the Series II, ultra-thin window type). During TEM the mean diameter of the electron beam was 2 nm. Metallographic samples were removed at the quarter-point location (midway between the centreline and the periphery of the component). All metallographic test samples were removed transverse to the joint interface and polished and etched in a mixture of perchloric acid and glacial acetic acid. During TEM microscopy, 1 mm thick samples were cut and wet-polished to 0.2 mm thickness. Finally, the TEM foils were prepared by dimpling and ion milling using argon. XRD analysis was carried out using a Rigaku Rint 1200 series device with a copper target.

3. Results

3.1. Joint mechanical properties

Fig. 3 illustrates the effect of holding time at 823 K and 873 K on the tensile strength of Ti/h.p. Al and Ti/c.p. Al joints. Holding times up to 5 h at 823 K and 1 h at 873 K had a negligible effect on the tensile strength of Ti/h.p. Al joints and all test samples failed in the aluminium substrate. However, increasing the holding time at 823 K and 873 K markedly decreased joint strength properties of Ti/h.p. Al joints. For example, the tensile strength of Ti/h.p. Al joints decreased drastically for heating times of 10 h at 823 K and 2.5 h at 873 K and all test samples failed at the joint interface. In Ti/c.p. Al joints, the tensile strength properties were unaffected by holding times up to 1 h at 873 K and the tensile strength decreased markedly when the holding time was 50 h at 873 K.

Fig. 4 shows the influence of holding time at 823 and 873 K on the bend test properties. The bend test results mirrored those found during tensile testing, e.g.

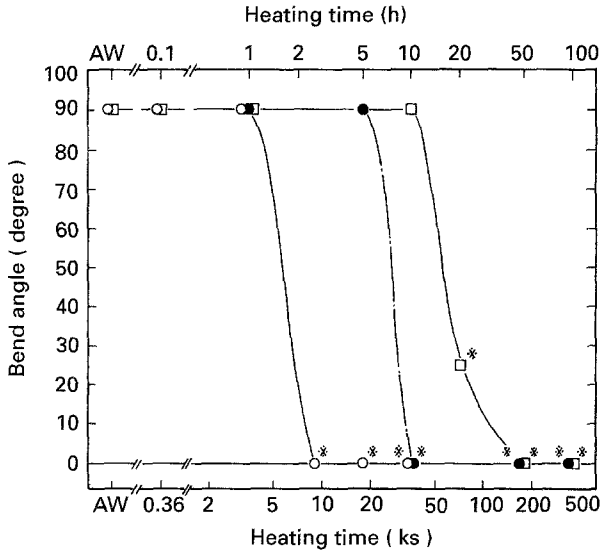

Figure 4 Effect of holding time on the bend test properties of Ti/h.p. Al and Ti/c.p. Al joints. □ 873 K, Ti/c.p. Al; ○ 873 K, Ti/h.p. Al; ● 823 K, Ti/h.p. Al; * fractured at interface.

Ti/h.p. Al joints produced very low bend angles when the holding time was 10 h at 823 K and 2.5 h at 873 K and test specimen failure occurred at the joint interface region. In Ti/c.p. Al joints, the bend angle only decreased markedly when the holding time at 873 K exceeded 10 h. It follows that the responses of the Ti/h.p. Al joints and Ti/c.p. Al joints to heat treatment were quite different. In particular, the application of lower heat-treatment temperatures and shorter holding times markedly decreased the tensile strength and bend testing properties of Ti/h.p. Al joints. It will be shown later that the different responses to post weld heat-treatment were the result of differences in the rate of growth of Al_3Ti intermetallic layer at the joint interface.

3.2. Metallurgical examination

Fig. 5 shows optical micrographs of the bondline region in as-welded and post-weld heat-treated (873 K for 2.5 h) Ti/h.p. Al joints. Post-weld heat treatment allowed recrystallization of aluminium material adjacent to the joint interface. Fig. 6 shows the XRD results for the failure region of a sample heat-treated at 873 K for 10 h (this test specimen failed at the joint interface during tensile testing). Diffraction lines for aluminium, titanium and Al_3Ti are clearly apparent. It follows that failure of heat-treated joints during tensile testing was associated with fracture through the Al_3Ti intermetallic layer at the joint interface.

Fig. 7(a) shows a bright field transmission electron micrograph of an as-welded Ti/h.p. Al joint and the distributions of titanium, aluminium and silicon at the joint interface. Negligible interdiffusion was observed in the as-welded sample. However, a 1.2 μm wide interdiffused region was observed at the joint interface when the Ti/h.p. Al joint was heat-treated at 873 K for 0.1 h (see Fig.7(b)). The plateau region at the joint interface had the composition 25 at %Ti and 75 at % Al

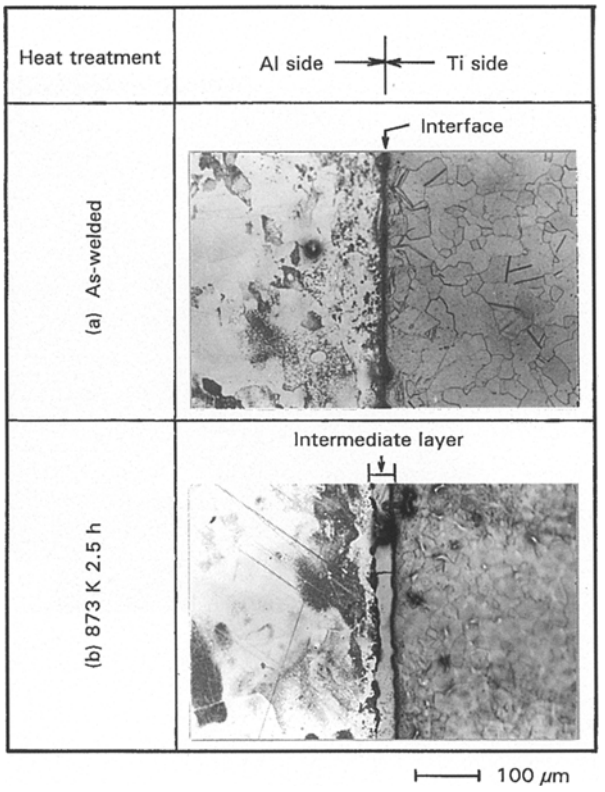

Figure 5 Optical micrographs of Ti/h.p. Al joints in the as-welded and heat-treated (873 K for 2.5 h) conditions.

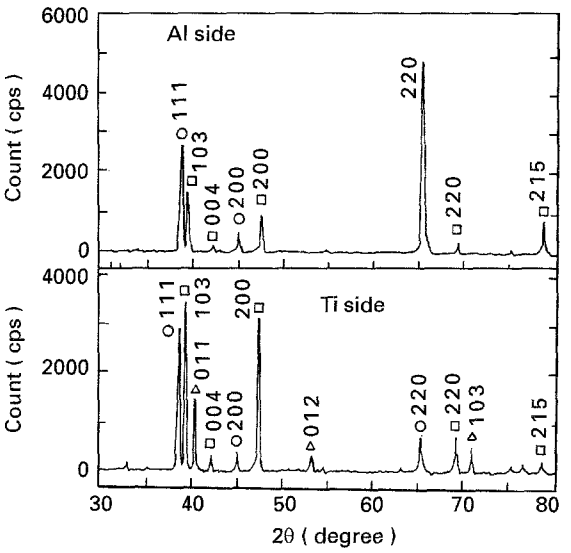

Figure 6 XRD analysis of the fracture surface region of a Ti/h.p. Al joint heat-treated at 873 K for 100 h. ○ Al; □ Al_3Ti; △ Ti.

and indicates Al_3Ti formation [6]. Small crystals were also apparent at the joint interface.

Fig. 8(a) and (b) show bright field transmission electron micrographs of as-welded and post-weld heat-treated Ti/c.p. Al joints. Again, there was negligible interdiffusion in the as-welded joint and the composition of the plateau region in the heat-treated joint was consistent with Al_3Ti formation at the joint interface. However, in the Ti/c.p. Al joint, the plateau region contained approximately 5 at % Si and more than 20 at % Si concentrated in the region immediately

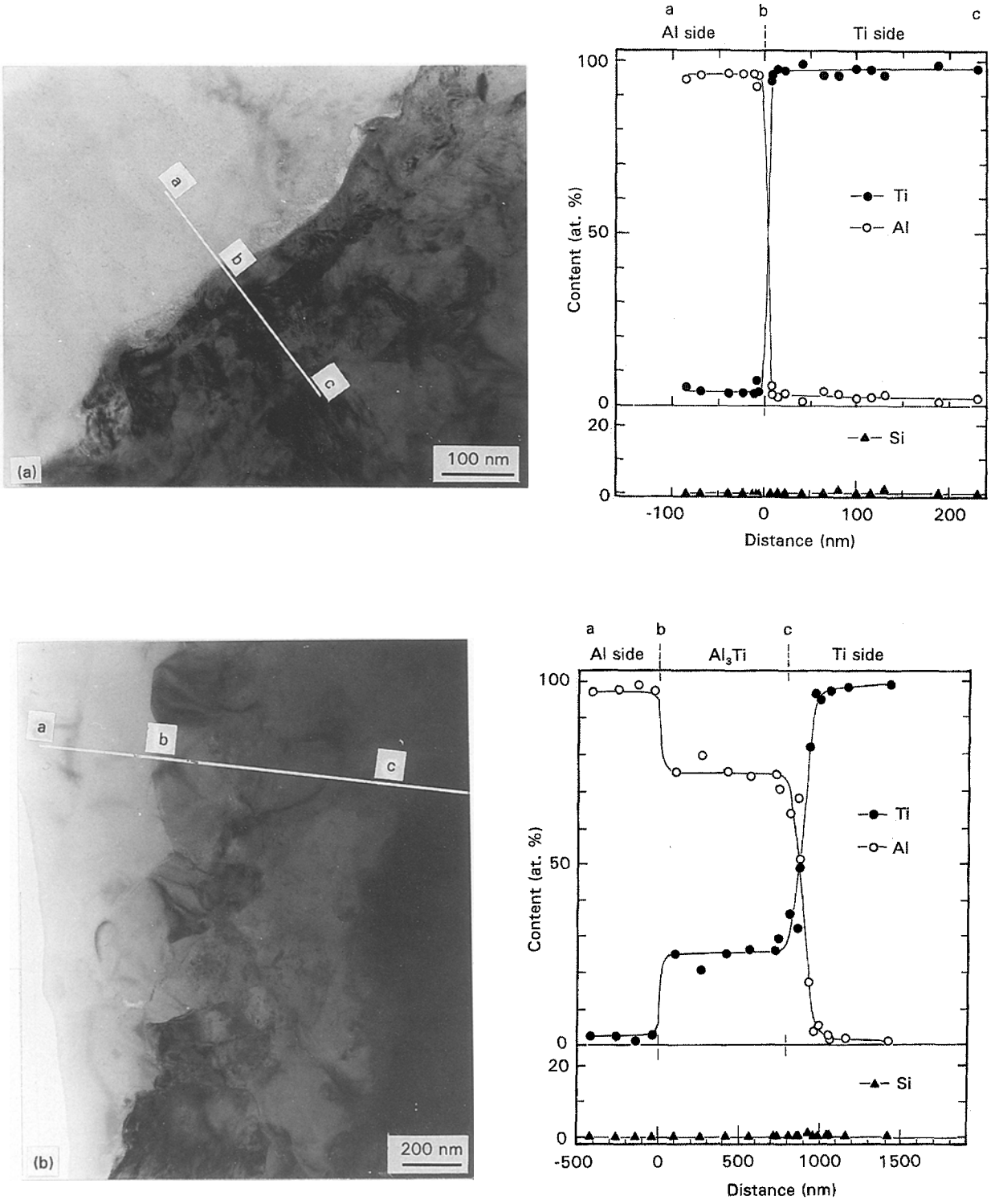

Figure 7 TEM of the joint interface region in (a) an as-welded Ti/h.p. Al joint and (b) a heat-treated (873 K 0.1 h) Ti/h.p. Al joint.

adjacent to the titanium substrate (see Fig. 8(b)). It is worth noting that there was no evidence of iron or copper segregation at the joint interface in heat-treated Ti/c.p. Al joints.

It is also worth pointing out that Fig. 7 and 8 consider holding times of 0.1 h and 10 h at 873 K respectively. The width of the intermediate layer in the Ti/h.p. joints heat-treated at 873 K for 1 h was too large to prepare a satisfactory TEM specimen. This occurred because the Al_3Ti intermetallic layer formed at a much faster rate in the heat-treated Ti/h.p. Al joint.

4. Discussion

Fig. 9 shows Arrhenius plots relating holding time, temperature and intermediate layer width in heat-treated Ti/c.p. Al and Ti/h.p. Al joints. When 1.5 μm and 5 μm wide intermediate layers were produced at the joint interface, the activation energies for intermediate layer formation were 207 kJ mol^{-1} (in the Ti/h.p. Al joint) and 193 kJ mol^{-1} (in the T/c.p. Al joint). In this connection, Enjo and Ikeuchi [7] indicated an activation energy of 195 kJ mol^{-1} for intermediate layer formation in diffusion welded Ti/c.p. Al joints. In Ti/c.p. Al joints, the increased concentration of silicon

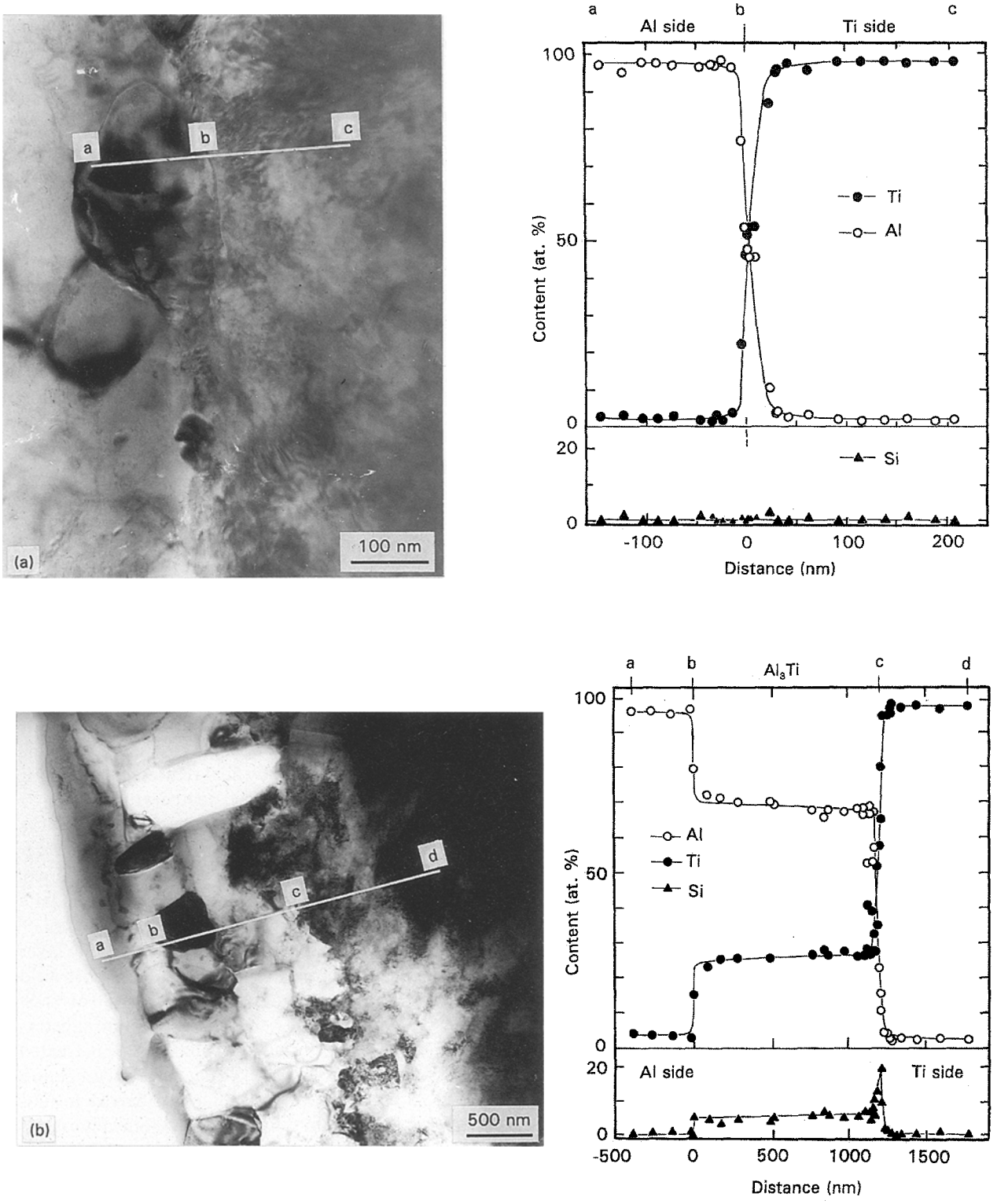

Figure 8 TEM of the joint interface region in (a) an as-welded Ti/c.p. Al joint and (b) a heat-treated (873 K 1 h) Ti/c.p. Al joint.

at the joint interface results from the difference in the diffusion rates of aluminium and of silicon in the α-titanium matrix. Nakajima and Koiwa [8] have indicated that the diffusion rate of silicon is three orders of magnitude higher than aluminium in α-titanium. Enjo and Ikeuchi [7] also observed that silicon concentrated at the joint interface region in diffusion-welded Ti/c.p. Al joints but did not confirm the exact location of the silicon segregation.

Although failure of heat-treated Ti/h.p. Al and Ti/c.p. Al joints was associated with Al_3Ti formation at the joint interface (see Figs 6 and 7), the responses of these dissimilar joints to post-weld heat treatment were quite different. In particular, the tensile strength and bend testing properties of Ti/h.p. Al joints were markedly decreased when lower post-weld heat treatment temperatures and shorter holding times were applied (see Figs 3 and 4). It is suggested that residual silicon in the c.p. Al substrate decreases the growth rate of the Al_3Ti intermetallic layer at the joint interface during post-weld heat-treatment. Segregation of silicon in the region between the titanium matrix and

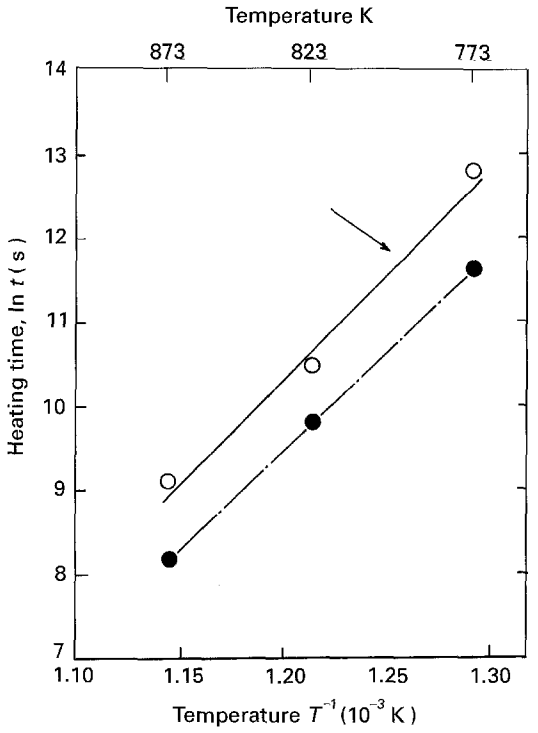

Figure 9 Holding time/temperature relations for the formation of a 5-μm thick intermediate layer in a Ti/h.p. Al joint (○, $Q = 207\ kJ\ mol^{-1}$) and for the formation of a 1.5-μm thick intermediate layer in a Ti/c.p. Al joint (●, $Q = 193\ kJ\ mol^{-1}$).

TABLE I Chemical compositions of the titanium and aluminium substrates

Chemical composition, (mass %)					Tensile properties	
(a) Titanium substrate						
H	O	N	Si	Ti	Tensile strength (MPa)	Elasticity (%)
0.007	0.135	0.006	<0.01	balance	497	27
(b) h.p. Aluminium substrate						
Si	Fe	Cu	Al			
6×10^{-4}	9×10^{-4}	37×10^{-4}	balance		52	50
(c) c.p. Aluminium substrate						
Si	Fe	Cu	Al			
0.12	0.54	0.13	bal.		92	47

the Al_3Ti layer retards growth of the intermetallic layer by acting as a barrier to titanium and aluminium diffusion. In this connection, alloy element segregation at the joint interface has also been associated with retardation of growth of the intermetallic layer in dissimilar friction joints between stainless steel and aluminium alloys containing copper [5]. It was also suggested that copper segregation enhanced joint strength by reducing the elastic strain at the joint interface (because of the smaller diameter of the copper atom compared to aluminium and iron atoms). It is worth noting that the c.p. Al substrate used in the present study contained 0.13 wt %Cu and 0.54 wt % Fe (see Table I). However, no segregation of copper or iron was detected at the joint interface (see Fig. 8(b)).

It has already been observed that dissimilar joint mechanical properties are drastically decreased when the thickness of the intermediate layer formed at the joint interface exceeds a critical value [3, 5, 9]. For example, a critical intermediate layer thickness of 1 μm/2 μm was indicated during dissimilar joining of AISI 304L stainless steel and titanium [3], niobium and Armco iron [9] and during pressure-welding of dissimilar materials [5]. In the present study, the tensile strength and bend test properties of both Ti/h.p. and Ti/c.p. joints markedly decreased when the intermediate layer width exceeded 10 μm. The critical intermediate layer width in Ti/c.p. Al and Ti/h.p. Al joints is consequently much larger than the 1 μm/2 μm value indicated in other investigations. It follows that the critical intermediate layer width depends on the mechanical properties of the intermetallic phase formed at the joint interface and of the substrate combination considered.

5. Conclusions

The influence of post-weld heat treatment and of residual silicon in aluminium on the mechanical properties of dissimilar friction joints between titanium and aluminium was investigated. The following conclusions were reached:

1. Although joint tensile strength and bend test properties were drastically reduced following post-weld heat treatment, the responses of Ti/h.p. Al and Ti/c.p. Al joints were quite different. The tensile strength and bend test properties of Ti/h.p. Al joints were markedly decreased by heat treatments involving shorter holding times at lower temperatures.
2. Joint failure in post-weld heat-treated joints was associated with Al_3Ti formation at the bondline region. Joint failure in dissimilar Ti/h.p. Al and Ti/c.p. Al joints occurred when the width of the intermediate layer formed at the joint interface exceeded 10 μm.
3. The growth rate of the Al_3Ti intermetallic layer at the joint interface was much faster in post-weld heat-treated Ti/h.p. Al joints. More than 20 at % Si segregated in the region between the titanium substrate and the Al_3Ti intermetallic phase in heat-treated Ti/c.p. Al joints. It is suggested that silicon segregation retards Al_3Ti formation by acting as a barrier to titanium and aluminium diffusion at the joint interface.

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