Fracture behaviour of diffusion bonded bimaterial Ti–Al joints

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Fracture behaviour of diffusion bonded bimaterial Ti–Al joints

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Failure modes of constrained metal foils between two elastic solids are rather different from those in the unconstrained condition. If the interface adhesion is strong between materials, a lower strength thin metal (plastic) foil between two much higher strength metals (elastic) can undergo substantial plastic deformation and fail with high triaxiality induced ductile fracture. Experiments have been conducted to explore the modes of failure and the factors governing fracture in such a constrained metal interlayer. In the present work, the effects of soft interlayer thickness and brittle reaction layer on the fracture behaviour of four point bend specimens have been investigated. A series of solid state diffusion bonds were produced between 25 × 25 mm section titanium bars using pure aluminium foils of different thickness (50, 457, 914, and 2000 μm) as the soft constrained interlayer. All four point bend specimens containing an ∼2 μm thick intermetallic reaction layer $TiAl_3$ between the titanium and aluminium failed in ductile fracture mode within the soft aluminium interlayer next to the interface. A number of void formations were observed ahead of the crack tip next to the interface. No evidence of interface debonding was observed. However, the specimens containing an 8 μm thick $TiAl_3$ layer failed by brittle fracture along the interface between the titanium substrate and the $TiAl_3$ layer. It was found that decreasing the soft interlayer thickness from 2000 to 457 μm increased the load carrying capacity and decreased the fracture toughness caused by constrained plastic deformation (high triaxiality) of the interlayer.

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INTRODUCTION

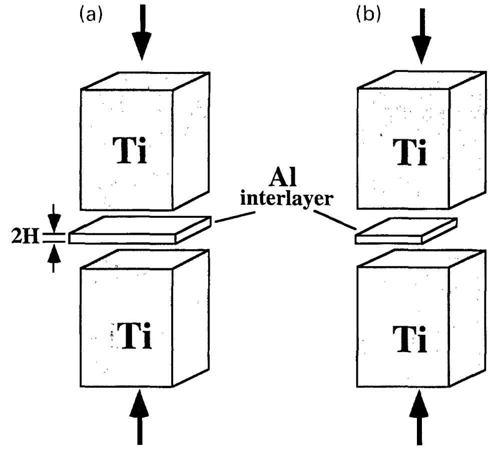
A welded or bonded bimaterial joint is a common feature in many structural components. Diffusion bonding is a relatively inexpensive solid state process which does not melt the base materials to be joined, thus avoiding undesirable phase transformations and solidification cracking. Furthermore, it offers good potential for joining dissimilar materials and advanced materials such as Ti–Al intermetallics; fusion welding processes are not applicable for joining these materials owing to their susceptibility to solid state cracking resulting from their poor room temperature ductility.[1,2] An interface between dissimilar materials can be a critical location in a bimaterial component. In contrast with problems of cracking found in homogeneous bodies, the bimaterial interface quality links joint performance with joining process variables and mechanical properties of the two materials. Therefore, a complete characterisation of an interface fracture process requires interfacial fracture toughness data over the full range of joining process variables, interlayer thicknesses, and fracture mode combinations.

The fracture mechanics of elastic/plastic bimaterial interfaces (one of the constituents deforms plastically) has recently been the focus of intense research to improve the understanding of the fracture process occurring at bimaterial interfaces.[3–16] Recently, Kim *et al.*[17,18] have conducted numerical and slip line field analysis on bimaterial single edge bend and centre cracked tension configurations. Their results showed the development of ∼38% higher triaxiality at the interface compared with that of homogeneous specimens.

For experimental evaluation of the interfacial fracture toughness of similar or dissimilar joints, sandwich test specimens have generally been used.[6] These specimens contain a very thin interlayer (material 1) which is sandwiched between the two solid blocks (material 2) comprising the bulk of the specimen. Such specimens can be produced by the solid state diffusion bonding (DB) process. A sandwich structure comprising a softer metal layer between higher strength metal blocks may fail in a variety of ways, including brittle debonding of interfaces and ductile rupture of the interlayer. The fracture process of such joints depends on the level of elastic/plastic mismatch, interlayer thickness $2H$, and uncracked ligament size $2H/(W-a)$ where W is the specimen width and a is the initial crack length.

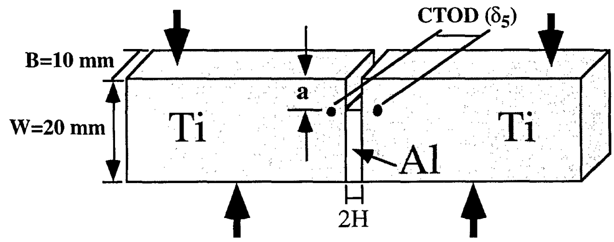
The presence of a brittle reaction layer along the constrained metal foil will obviously complicate the failure mode of the system (material 2–material 1–material 2) considered. An intermetallic $TiAl_3$ reaction layer forms during the solid state bonding of titanium to aluminium.[3] The formation of such brittle intermetallic phases next to the soft interlayer can introduce an additional failure mechanism (brittle interface debonding). In sandwich structures with high toughness interlayers the failure is expected to take place in the soft interlayer in the vicinity of the interface, provided that brittle interface debonding is avoided (strong interface adhesion).[12,14] The presence of a soft, thin aluminium interlayer (plastic) between two high strength titanium bulk material blocks (elastic) can present an analogy to metal–ceramic bonds. Therefore, it can be expected that the Ti–Al–Ti system will fail by a fully ductile mode (nucleation, growth, and coalescence of voids) owing to fully contained plastic strain within the soft aluminium interlayer in the vicinity of the interface, provided that the intermetallic reaction layer between titanium and the soft aluminium interlayer is not excessively thick or the interlayer is not extremely thin.

In the present study, interfacial fracture toughness and tension tests were carried out on Ti–Al–Ti specimens produced using the DB process. This paper reports only the mechanical fracture toughness aspects and the effect of the presence of brittle intermetallic $TiAl_3$ on the fracture behaviour of the constrained layer in diffusion bonded Ti–Al–Ti joints. The competition of the various failure modes,



a tensile; *b* four point bending

1 Schematic illustration of DB of Ti–Al–Ti system for given specimens



2 Four point bending specimen configuration extracted from diffusion bonded sandwiched Ti–Al–Ti specimens: $a/W = 0{\cdot}35$

therefore, can be studied by a systematic analysis of this system. Microstructural characterisation of the diffusion bonded interfaces, prebonding treatments, and bonding parameters were reported earlier.[3]

EXPERIMENTAL PROCEDURE

Titanium blanks of purity 99·5% and size 25 × 25 × 55 mm were diffusion bonded with pure aluminium foils of purity 99·0%. These foils had $2H$ values of 50, 457, 914, and 2000 µm before bonding (Fig. 1). Some creep, particularly in thicker foils, occurred during bonding and caused a slight decrease in the foil thickness. However, it was difficult to measure the amount of creep accurately because of interdiffusion also occurring during bonding. In addition to various aluminium foil thicknesses, several different temperature, pressure, and bonding time schedules were investigated (Table 1). Tensile tests of pure titanium and aluminium bars (having similar purity levels to the aluminium foils used) were carried out to determine the strength mismatch ratio between titanium and the constrained aluminium interlayer. An average of four tests was taken to determine the yield strengths σ_Y and ultimate tensile strengths σ_U of pure aluminium and titanium. Fracture toughness tests using fatigue precracked four point bending specimens of homogeneous titanium and aluminium bulk materials were also conducted to determine their bulk toughness values. Circular section tensile specimens (8 mm in diameter) were extracted from the bonds, which were produced by placing aluminium foil over the entire cross-section (Fig. 1*a*), and tested at room temperature.

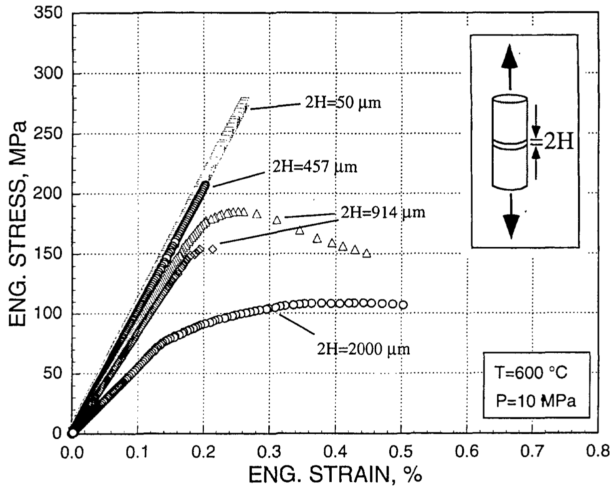
To produce four point bend specimens, thin aluminium foils were placed on a reduced cross-sectional area of ~19 × 24 mm between the titanium surfaces to be bonded in such a way as to obtain specimens with pre-existing notches, as shown in Fig. 1*b*. The diffusion bonds produced were cut longitudinally into two pieces to extract two, four point bending specimens ($a/W = 0{\cdot}35$) of thickness B 10 mm (producing, in fact, $B \times 2B$ specimens) as shown in Fig. 2. The specimens were not fatigue precracked and were tested at room temperature. Crack tip opening displacement (CTOD) values were directly measured using δ_5 (developed at GKSS) clip on gauges at the notch tip over a gauge length of 5 mm (Fig. 2). Additionally, load, displacement, and crack mouth opening displacement (CMOD) values were recorded during testing.

The specimens sectioned before and after tension and bend tests were investigated by optical and scanning electron microscopy to determine the quality of the bonds, fracture behaviour, and details of the fracture initiation and propagation at the interface.

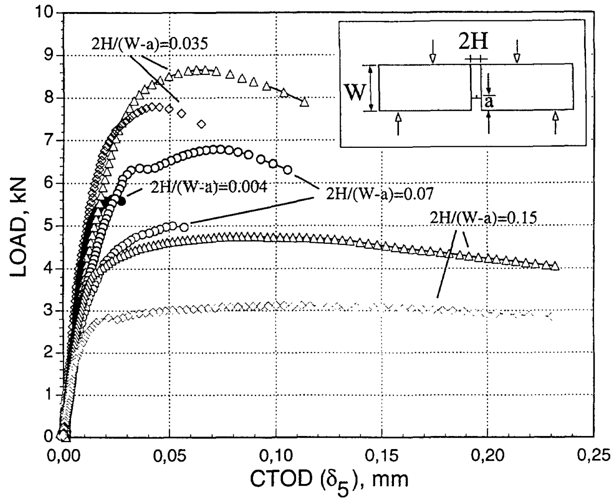
Table 1 Diffusion bonding parameters and mechanical properties* for tensile and four point bend specimens of Ti–Al–Ti system

Bonding parameters	Foil thickness, µm	Tensile strength, MPa	CTOD (δ_{5m}), mm	Remarks for bend test
600°C, 10 MPa, 1 h	50	215·07	0·006	Brittle fracture at interface
		112·91	0·008	(bonding time not optimum)
600°C, 10 MPa, 5 h	50	277·37	0·023	Ductile fracture
		276·50	0·027	
600°C, 10 MPa, 1 h	457	172·17	0·067	Ductile fracture
		128·73	0·041	
600°C, 10 MPa, 3 h	457	79·50	0·067	Ductile fracture
		185·26	0·055	
600°C, 10 MPa, 5 h	457	207·17	0·065	Ductile fracture
		213·12	0·048	
600°C, 10 MPa, 30 h	457	220·05	0·017	Brittle debonding at $Ti/TiAl_3$
		...	0·025	interface
600°C, 10 MPa, 1 h	914	107·47	0·070	Ductile fracture
		160·97	0·052	
600°C, 10 MPa, 5 h	914	185·37	0·025	Ductile fracture
		153·92	0·098	
600°C, 10 MPa, 1 h	2000	93·65	0·090	Ductile fracture
		108·84	0·100	

* Two identical specimens tested for each set of conditions.



3 Stress–strain curves for specimens bonded with different soft interlayer thicknesses showing increase of strength with decreasing soft interlayer thickness



4 Load *v.* CTOD (δ_5) curves for specimens bonded with different interlayer thicknesses showing effect of $2H/(W-a)$ ratio

RESULTS AND DISCUSSION

The mechanical properties and fracture toughness values for the bulk titanium and aluminium are given in Table 2. The strength mismatch of the Ti–Al–Ti system represents an extreme undermatching case with M, the mismatch ratio, having a value of 0·19. The CTOD values at maximum load δ_{5m} of bulk titanium and aluminium were 0·54 and 3·36 mm, respectively. Tensile test results for the bonds are summarised in Table 1, which also gives CTOD (δ_{5m}) values for the joints corresponding to the point of instability or maximum load attained for each four point bend specimen. Tensile results, especially for shorter bonding times (1–3 h), reflect some scatter, which can be attributed to inhomogeneous bond quality across the cross-section of large specimens (Fig. 1). Figure 3 shows the stress–strain curves obtained from Ti–Al–Ti circular section tensile specimens with varying aluminium $2H$ values. Bonding time was 5 h except for 2000 µm aluminium foil, which was bonded for 1 h to avoid the excessive creep associated with longer bond times. This result for the joint made using a thick interlayer with 1 h bonding time was included for the sake of clear comparison to demonstrate the effect of soft interlayer thickness on the deformation behaviour of the joints. A previous study[3] has shown that bonding times from 1 to 5 h did not significantly change the deformation behaviour of the joints (*see* Figs. 5 and 8 in Ref. 3). The decreasing soft aluminium interlayer thickness clearly increases the strength and drastically decreases the ductility owing to the constrained plastic deformation of the interlayer. The strengths of the bonds significantly exceed the strength of the soft aluminium interlayer as a result of the constrained plasticity, except in the case of a thicker foil, i.e. one 2000 µm thick. The soft aluminium interlayer is restrained by the non-deforming titanium, which means that the effective yield strength of the aluminium is increased significantly; this is also the case (contact strengthening) in brazed joints.[19] Failure occurred by ductile rupture in the aluminium next to the bond interface. No interface debonding could be detected and, for thinner layers, the stress–strain curves exhibited no detectable yield point (Fig. 3).

Table 2 Mechanical properties* and fracture toughness values for bulk Ti and Al

Material	σ_Y, MPa	σ_U, MPa	CTOD (δ_{5m}), mm
Bulk Ti	595	690	0·54
Bulk Al	115	125	3·36

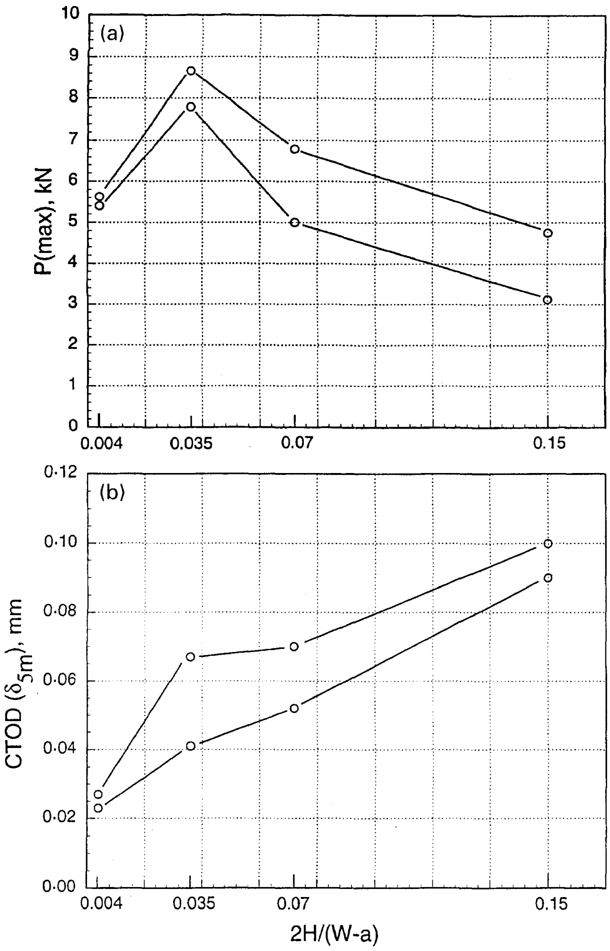
* Mismatch ratio $M=\sigma_Y(\mathrm{Al})/\sigma_Y(\mathrm{Ti})=0{\cdot}19$.

The formation of an intermetallic reaction layer between the titanium and aluminium substrates and the effect of the bonding conditions on the mechanical properties were reported in more detail earlier.[3,4] In the present paper, the effect of thickness of the soft aluminium interlayer and the brittle intermetallic reaction layer ($TiAl_3$) on the fracture behaviour of the sandwich Ti–Al–Ti structures will be discussed.

Fracture behaviour of single edge notched bend specimens

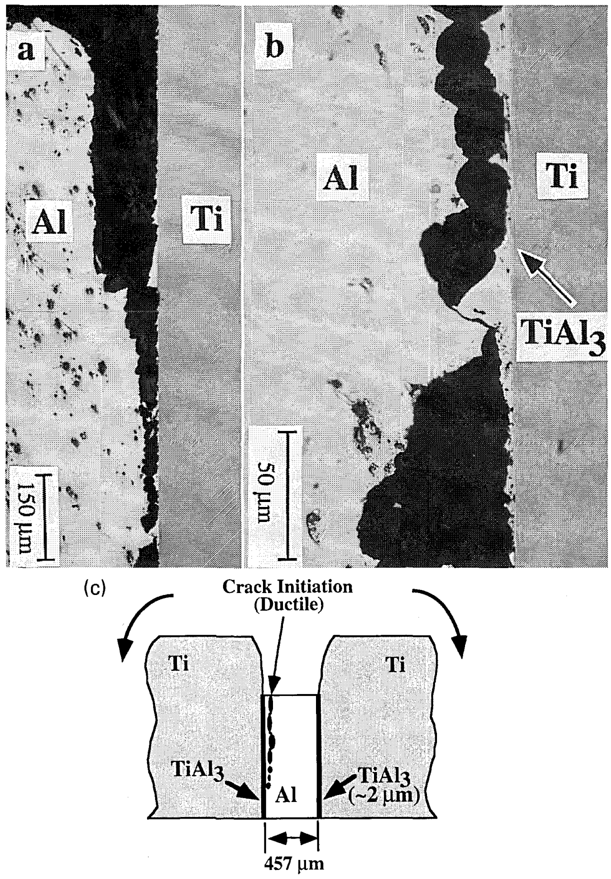
Figure 4 shows load *v.* CTOD (δ_5) curves for single edge notched bend (SENB) specimens bonded for 1 h (except the 50 µm aluminium foil specimen, which was bonded for 5 h). For each condition two SENB specimens were tested and any difference observed between two specimens of identical condition resulted from small variations in the a/W ratio. It is clearly seen that a decrease in the soft aluminium interlayer thickness from 2000 to 457 µm results in an increase in the load carrying capacity of the specimens at the expense of toughness. A further decrease in the soft aluminium interlayer thickness from 457 to 50 µm caused a sudden decrease in load carrying capacity (Fig. 4). The bonding time of 5 h was used for the 50 µm aluminium foil to produce a strong bond since the joint produced with a 1 h bonding time did not provide sound joint performance owing to a stable oxide layer, as reported earlier.[3] No difference in interface microstructure or thickness of the reaction layer was observed between the 1 and 5 h bonded specimens.

Figure 5*a* illustrates the relationship between the ratio $2H/(W-a)$ and maximum load levels obtained by the SENB specimens in Fig. 4. Both Figs. 4 and 5*a* clearly show that the maximum load increases to a certain level with decreasing $2H/(W-a)$ ratio and then the occurrence of sudden instability (ductility) decreases the maximum load level that can be achieved. To determine the exact position of the transition point in the curve, further bonds will be made for testing in the future using soft aluminium interlayers with different thicknesses between 457 and 50 µm. Fracture surface examination of all specimens in the present study (including $2H/(W-a)=0{\cdot}004$) showed a fully ductile fracture mode of failure (Fig. 6). Clearly, there is no change of fracture mode of specimens with decreasing interlayer



a on maximum load attained; *b* on CTOD (δ_{5m}) values: CTOD (δ_{5m}) of bulk Al is 3·36 mm

5 Effect of $2H/(W-a)$ ratio for Ti–Al–Ti bend test specimens



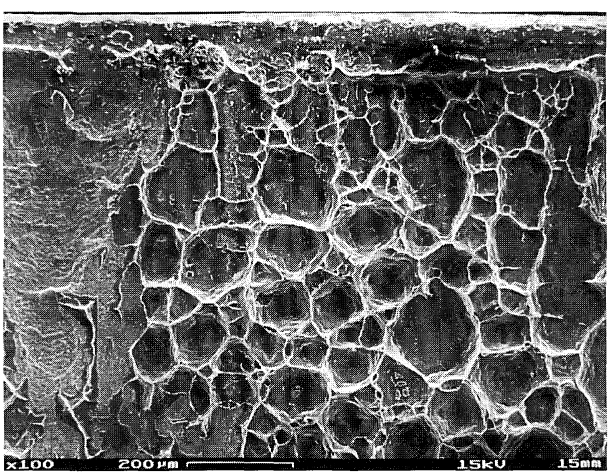
a optical image, low magnification, 914 µm; *b* optical image, high magnification, 914 µm; *c* schematic diagram, 457 µm

7 Sections of SENB specimen with given thickness soft Al interlayer showing ductile fracture process by void nucleation and coalescence in soft interlayer next to interface: no evidence of brittle interface debonding despite presence of brittle $TiAl_3$ reaction layer

thickness. However, the fracture stress is strongly influenced by the thickness of the interlayer.

Figure 5*b* shows the relationship between the ratio $2H/(W-a)$ and CTOD. The CTOD (δ_{5m}) value decreases with decreasing soft interlayer thickness.

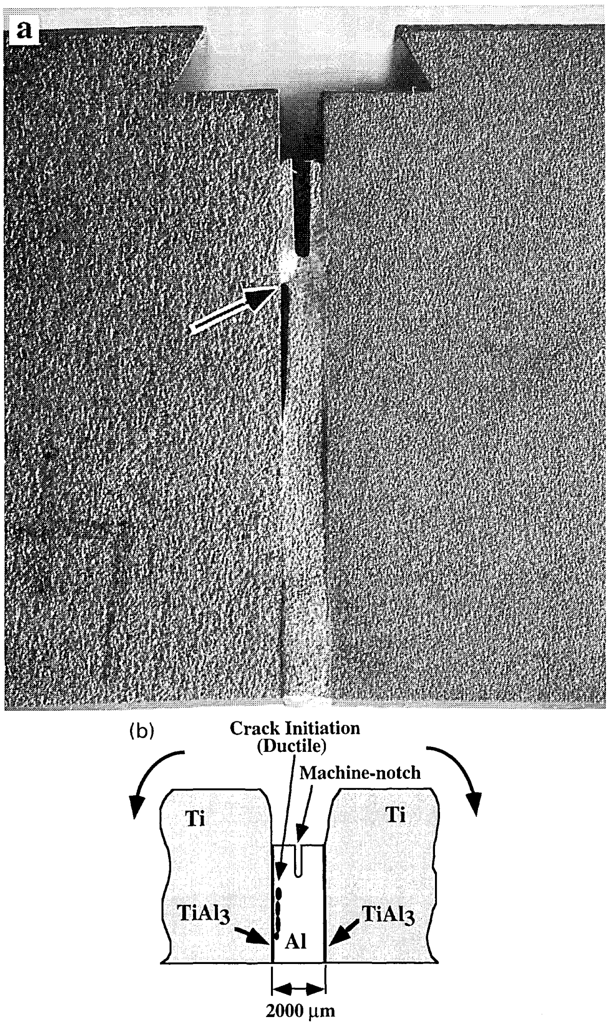
All the bonds made at 600°C and 10 MPa exhibited a ductile mode of failure, fracture occurring within the soft aluminium interlayer near the interface. In these bonds, fracture proceeded by the nucleation, growth, and coalescence of voids in the soft aluminium interlayer next to the interface under constrained plasticity. Figure 7*a* and *b* illustrates crack initiation and extension by ductile void growth and coalescence in the aluminium next to the bond interface. No brittle debonding at the interface is evident. It is important to note that the formation of voids and coalescence occurred entirely within the weaker aluminium interlayer as can be seen clearly in Fig. 7*b*, indicating that the interface bond is strong enough to allow the soft aluminium interlayer to undergo substantial plastic deformation. The ductile failure mechanism observed for SENB specimens is schematically shown in Fig. 7*c*, which illustrates the observed fracture mode presented in Fig. 7*a* and *b*.



6 Fracture surface of specimen with 50 µm soft interlayer showing ductile fracture topography

To examine void nucleation, growth, and coalescence behaviour in detail, a very fine spark erosion machine notch was introduced into the soft aluminium interlayer ($2H =$ 2000 µm). It was found that, even though the machine notch was in the middle of the soft interlayer, the first void nucleation and crack initiation took place ahead of the crack tip within the soft interlayer next to the interface, as shown in Fig. 8*a* and *b*. The cavities that developed next to the interface did not connect with the crack tip. Obviously, such a failure mode is rather different from that for a crack in a homogeneous bulk material where voids nucleate and grow at and ahead of the crack tip within distances of the order of crack tip opening.

Determination of the exact distance of void nucleation from the crack tip with respect to interlayer thickness is currently in progress. The observation of crack initiation



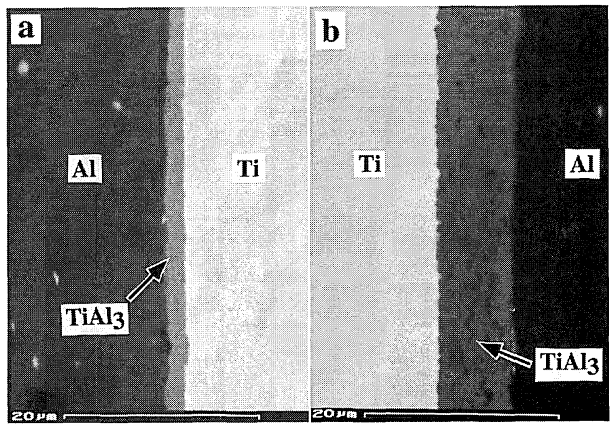
a optical image; *b* schematic diagram

8 Failure mechanism in Ti–Al–Ti sandwich joint with very fine machine notch: crack initiates within soft Al interlayer next to interface rather than at tip of notch

ahead of the crack tip indicates that the peak stress does not develop at the crack tip itself under such a constrained plastic condition of the soft interlayer. A finite element analysis by Varias *et al.*[12] revealed that the hydrostatic stress in a metal foil (metal–ceramic system) can easily exceed 5 times its yield stress. Such high stresses at a distance of several foil thicknesses ahead of the crack tip can cause void formation at those locations. These experimental observations have been confirmed by various numerical analyses of metal–ceramic systems.[5,7,9–12] The formation of a void nucleation and growth mechanism will be restricted with decreasing interlayer thickness if the interlayer thickness approaches the mean void spacing of the interlayer. In other words, the spacing between two void nucleation sites becomes a dominating factor for fracture rather than the interlayer thickness.

Effect of brittle reaction layer thickness on fracture behaviour

To determine the effect of the brittle intermetallic layer thickness on the fracture behaviour, specimens were bonded for 5 and 30 h at 600°C and 10 MPa using a 457 μm aluminium interlayer. The variation in thickness of the brittle intermetallic $TiAl_3$ reaction layer with bonding times of 5 and 30 h is illustrated in Fig. 9*a* and *b* respectively. Figure 9*a* shows a brittle reaction layer of ~2 μm that was



a 2 μm; *b* 8 μm

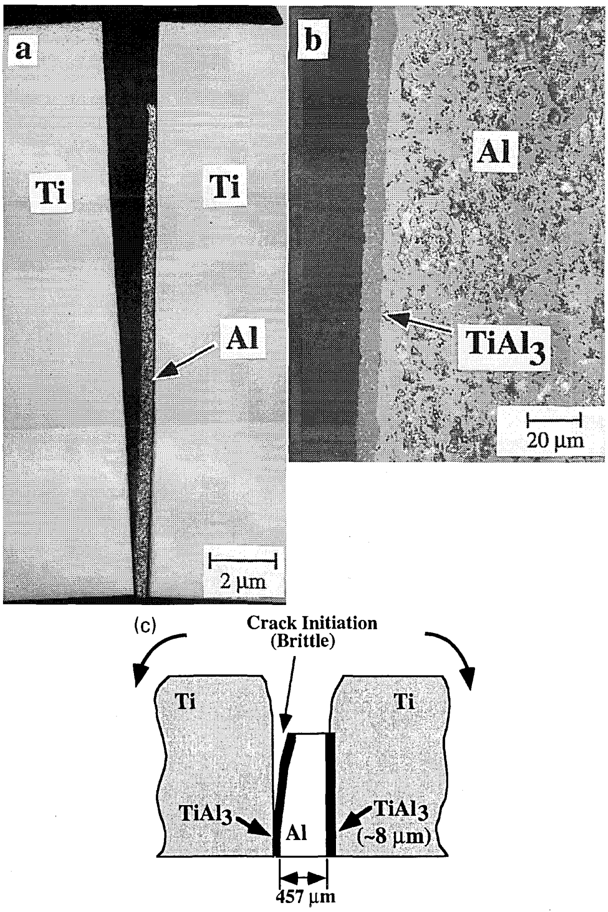
9 Backscattered electron images of bonds containing given thicknesses of $TiAl_3$ reaction layer: specimens diffusion bonded at 600°C and 10 MPa, Al foil thickness $2H=$ 457 μm

produced with a bonding time of 5 h, whereas Fig. 9*b* illustrates an intermetallic reaction layer of ~8 μm that was formed after a bonding time of 30 h. The effect of the 30 h bonding time on the flow stress of aluminium foil has not yet been studied.

The presence of a thin brittle layer of ~2 μm next to the soft aluminium interlayer apparently does not affect the ductile fracture process that takes place within the interlayer as shown in Fig. 7. No microcracks within the interphase, which are connected to the voids formed within the soft aluminium interlayer, were observed. Similarly, ductile fracture by void formation and coalescence was also observed on the metal side[5,14] of ceramic–metal dissimilar joints with or without a reaction layer on the ceramic side. However, in the present study, the presence of ~8 μm of brittle intermetallic $TiAl_3$ reaction layer promoted a brittle fracture process which took place along the interface between the titanium substrate and the intermetallic $TiAl_3$ reaction layer. The brittle fracture along the titanium/$TiAl_3$ interface is illustrated in Fig. 10*a* and *b*. No void formation was observed in the soft aluminium interlayer. The presence of two high strength phases of $TiAl_3$ between the bulk titanium and the aluminium foil may affect the stress state of the uncracked ligament depending on the relative thickness of this reaction layer. The brittle fracture process is also schematically shown in Fig. 10*c*. Brittle fracture along the interface in ceramic–metal joints without reaction product or interphases was also observed.[14]

The specimens containing a brittle intermetallic layer of ~8 μm did not display any ductility in four point bend and tension tests. The load *v.* CTOD curves for the bonds with ~2 and ~8 μm intermetallic $TiAl_3$ reaction layers are shown in Fig. 11. The presence of the 8 μm $TiAl_3$ reaction layer did, however, increase the load carrying capacity of the joint by possibly introducing additional constraint (owing to inherent limitations on slip) to the soft interlayer (Fig. 11). An increase of reaction layer thickness from 2 to 8 μm clearly changes the fracture mode of the specimens. The mere presence of a reaction layer obviously does not change the fracture mode, but it depends on the relative thickness of the reaction layer of the system studied. The level of stress state reached in the specimens containing an 8 μm $TiAl_3$ layer apparently was high enough to initiate brittle fracture at the brittle $TiAl_3$ phase. The stress relaxation effect of the soft aluminium interlayer was not enough to prevent brittle fracture in these specimens.

It should, however, be mentioned that the presence of residual stresses within the intermetallic reaction layer may possibly cause secondary microcracks within this brittle



a optical image, low magnification; *b* optical image, high magnification; *c* schematic diagram

10 Sections showing brittle fracture taking place at Ti/$TiAl_3$ interface in Ti–Al–Ti sandwich specimens bonded at 600°C and 10 MPa for 30 h: thickness of brittle $TiAl_3$ reaction layer is ~8 µm

layer during loading of the specimens, particularly in those with a thick reaction layer. Further work is needed to explain the reason for brittle fracture in specimens containing 8 µm of $TiAl_3$ intermetallic reaction layer.

CONCLUSIONS

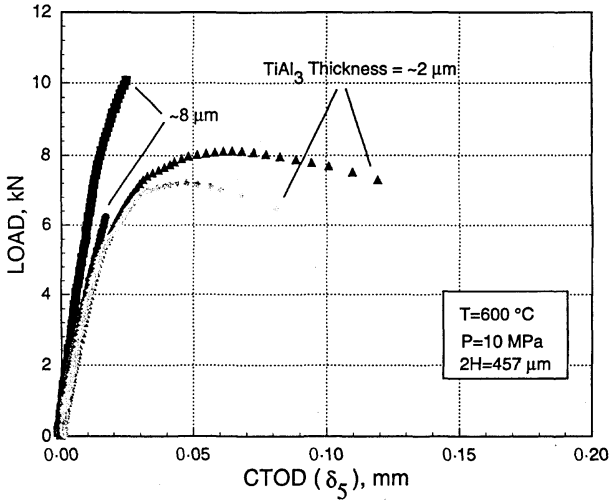
1. The solid state diffusion bonded Ti–Al–Ti system has been used to investigate the deformation and fracture process in a constrained ductile metal layer by considering the interlayer and brittle reaction layer thicknesses.

2. The optimum solid state diffusion bonding parameters to obtain a sound bond between pure titanium and a pure aluminium interlayer (provided that attention is given to adequate surface preparation before bonding) are pressure 10 MPa, temperature 600°C, and time 1–5 h.

3. It was found that the continuous interphase formed at the interface between titanium and aluminium is $TiAl_3$. The thickness of this brittle interphase is ~2 µm for the bonds obtained under the above conditions.

4. Bonds made with the longer time of 5 h exhibited higher bond strength values. The bond strength increases with decreasing constrained interlayer (aluminium foil) thickness at the expense of ductility.

5. All specimens failed by ductile fracture in the soft aluminium interlayer next to the bond interface with no evidence of interface debonding in bend specimens, including those made with a 50 µm aluminium foil. However, the bond made at 600°C and 10 MPa for 30 h using 457 µm aluminium foil, which contained an intermetallic layer of ~8 µm, displayed brittle debonding along the titanium/$TiAl_3$ interface.



11 Load *v.* CTOD (δ_5) curves for specimens bonded at 600°C and 10 MPa for 5 and 30 h using 457 µm Al interlayer showing effect of brittle reaction layer thickness

6. The load *v.* CTOD curves obtained for bend specimens exhibited increasing ultimate load level with decreasing soft aluminium interlayer thickness. The joint made using a 50 µm aluminium foil exhibited very limited deformation and hence the lowest CTOD values.

7. The presence of a brittle interphase ~2 µm in thickness did not result in any effect on the failure mode of the specimens, whereas that of an 8 µm brittle interphase resulted in brittle debonding along the titanium/$TiAl_3$ interface.

Further work is in progress to establish the micromechanism of the ductile fracture process and determine the *R* curve behaviour of this system using fatigue precracked shallow and deep notched bend specimens. Furthermore, numerical analysis of the Ti–Al–Ti system is in progress.

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