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Microstructural features of dissimilar MMC/AISI 304 stainless steel friction joints

C. PAN
Laboratory of Electron Microscopy, Wuhan Transportation University, Wuhan, Hubei 430063, People's Republic of China

L. HU
Department of Materials Science and Engineering, Huazhong, University of Science and Technology, Wuhan, Hubei, People's Republic of China

Z. LI, T. H. NORTH
Department of Metallurgy and Materials Science, University of Toronto, Toronto, Ontario, Canada M5S 1A4

The microstructure and mechanical properties of dissimilar friction joints between aluminium-based MMC and AISI 304 stainless steel base materials were investigated. The microstructural features which occur in the stainless steel substrate comprise deformation twinning, formation of a fine-grained dislocation substructure in austenite, and plastic deformation. The stainless steel substrate was plastically deformed in the region close to the mid-radius of the dissimilar joint. In a similar manner, 5–10 μm thick transition layers, comprising regions of locally plasticized MMC-base material, were formed close to the mid-radius location in dissimilar joints. The interlayer formed at the dissimilar joint interface comprised a mixture of oxide ($Fe(Al,Cr)_2O_4$ or $FeO(Al,Cr)_2O_3$) and $FeAl_3$ intermetallic phases. The notch tensile strength of dissimilar MMC/AISI 304 stainless steel joints increased when the rotational speed increased from 500 r.p.m. to 1000 r.p.m., and at higher rotation speeds, there was no effect on notch tensile strength properties.

1. Introduction

Aluminium-based metal matrix composites (MMC) have the inherent advantages of high specific strength, elastic moduli and wear resistance. General application of these MMC-base materials will depend to a large extent on the effectiveness of fabrication techniques which can join these composite materials to themselves and to other substrates such as stainless steel. With this in mind, the present paper examines the factors determining the metallurgical and mechanical properties of friction joints involving aluminium-based MMC and stainless steel base materials.

It has been suggested that the mechanical properties of dissimilar friction joints between aluminium and stainless steel are largely determined by intermetallic formation at the joint interface region [1–3]. For example, $FeAl_3$ and Fe_2Al_5 intermetallic phases were detected at the joint interface of aluminium/stainless steel joints following a heat treatment comprising 4 h at 773 K [3]. However, these heat-treatment conditions in themselves can provide the conditions for intermetallic formation at the joint interface. With this in mind, these experimental results cannot be taken as confirmation that intermetallic phases such as $FeAl_3$ and Fe_2Al_5 were actually present in as-welded aluminium/stainless steel joints. In this connection, Chernenko [4] and Yashan *et al.* [5] did not detect intermetallic formation at the joint interface in friction joints between aluminium and stainless steel. In addition, following detailed transmission electron microscopy, Fuji *et al.* [6] confirmed that as-welded friction joints between titanium and aluminium substrates were free of intermetallic phases while post-weld heat-treated joints contained Al_3Ti. Therefore, it is important to confirm the type of interlayer formed at the joint interface during dissimilar MMC/AISI 304 stainless steel friction joining.

Diverse views have been expressed concerning the role played by the harder (high flow stress at high temperature) substrate during dissimilar friction joining. Chernenko [4] suggested that some deformation of the harder (stainless steel) substrate must occur if satisfactory joint mechanical properties are to be obtained between aluminium and stainless steel substrates. When the difference in the flow strengths of the harder and softer substrates is too great, negligible deformation will occur in the harder substrate and relatively poor joint mechanical properties will result. However, Sada and Bahrani [7] ascribed an essentially passive role to the harder substrate, i.e. (a) the harder substrate will be completely covered with a coating of the softer base material during the initial stage in dissimilar friction joining, and (b) this coating will act as a stagnant layer so that the rubbing plane will be shifted axially towards the softer base material. With this in mind, the present work investigated the microstructural features of dissimilar MMC/AISI 304 stainless steel joints, with particular emphasis on the changes that occur in the stainless steel substrate in the region immediately adjacent to the bondline. A combination of scanning electron microscopy and transmission electron microscopy have been applied to fulfil this objective.

It has been suggested that intermetallic formation at the dissimilar Al/AISI 304 stainless steel joint interface is promoted when high rotational speed and low frictional pressure values are employed during dissimilar friction joining of aluminium and stainless steel substrates [1]. The lowest joint strength values were reported in joints produced using high rotational speeds. Higher rotational speed values were associated with the generation of higher temperatures at the dissimilar joint interface and with promotion of enhanced interdiffusion and growth of brittle intermetallic phases. The influence of rotational speed on joint microstructure and mechanical properties produced during dissimilar MMC/AISI 304 stainless steel friction joining have been examined.

2. Experimental procedure

The test materials comprised 19 mm diameter bars of alloy 6061/Al_2O_3 (W6A. 10A-T6) base material containing 10 vol% Al_2O_3 particles and AISI 304 austenitic stainless steel. The nominal chemical composition of the MMC base material was 0.28 wt% Cu, 0.6 wt% Si, 1 wt% Mg, 0.2 wt% Cr, balance aluminium.

Friction welding was carried out using a direct-drive friction welding machine. The influence of joining parameters on notch tensile strength was evaluated in tests where the rotational speed was varied from 500–2000 r.p.m. Table I shows the joining parameter settings in these tests. The dissimilar joint mechanical properties were investigated using notch tensile testing (the notch tensile test specimen design is shown in Fig. 1).

The microstructural features of the dissimilar joint interface were evaluated using scanning electron microscopy. The two-step etching procedure comprised (a) the stainless steel substrate being electro-etched using a solution of 10% chromic acid anhydride and water (using a voltage of 6 V and a current density of 0.1 A (cm^{-2}), and (b) the MMC substrate then being etched using a solution containing 1 ml HF, 2.5 ml HNO_3 and 95 ml water. Detailed microstructural observations were made in the following regions of the dissimilar joint: in the region between the specimen periphery and a location 4 mm from the joint periphery; in the region between a location 4 mm from the joint periphery and the sample centreline (this is close to the mid-radius location in the test joint); at the joint centreline.

During transmission electron microscopy, the test joints were sectioned using a spark-cutting machine to produce 0.5 mm thick specimens, and 3 mm diameter discs were then punched out at the bondline region. The test specimens were then mechanically thinned to a thickness of 50 μm prior to ion thinning and the TEM foils were then examined using a Hitachi 200 kV H-800 transmission electron microscope. The TEM test specimens were extracted from friction joints produced using the following conditions: 30 MPa friction pressure, 120 MPa forging pressure, 1500 r.p.m. rotational speed and a friction time of 4 s.

TABLE I Joining parameter settings

Friction pressure (MPa)	Rotation speed (r.p.m.)	Forging pressure (MPa)	Friction time (s)
60	500	120	4.5
60	1000	120	4.5
60	1500	120	4.5
60	2000	120	4.5

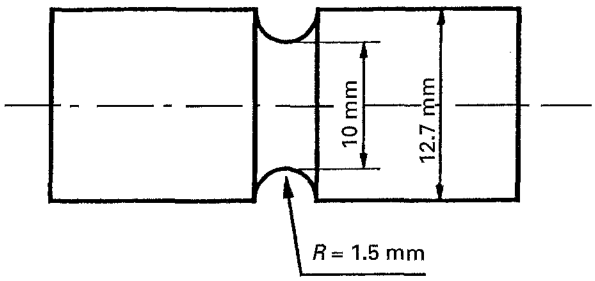


Figure 1 Notch tensile test specimen design.

3. Results and discussion

3.1. Joint interface microstructure

Fig. 2a–f show the microstructural features of the dissimilar MMC/AISI 304 stainless steel joint interface. Fig. 2b shows the as-received stainless steel microstructure comprising delta ferrite plates in an austenite matrix. Fig. 2c and d confirm the presence of deformation twinning and a fine-grained dislocation sub-structure of austenite in the region immediately adjacent to the bondline. Fig. 2e shows $FeAl_3$ interlayer formation at the dissimilar joint interface. Selected-area electron diffraction also confirmed retention of oxide material of composition $Fe(Al, Cr)_2O_4$ or $FeO(Al, Cr)_2O_3$ (see Fig. 3). Fig. 2f shows the microstructure of the as-received MMC substrate.

Fig. 4a–c show the microstructural changes observed in a traverse across the length of the dissimilar joint (in the stainless steel base material adjacent to the bondline). Deformation twinning was observed in the region between the joint periphery and a location 4 mm from the joint periphery (see Fig. 4a). The stainless steel substrate was plastically deformed in the region close to the mid-radius of the dissimilar joint (see Fig. 4b). At the joint centreline, there was no evidence of deformation twinning, nor in the stainless steel substrate (see Fig. 4c).

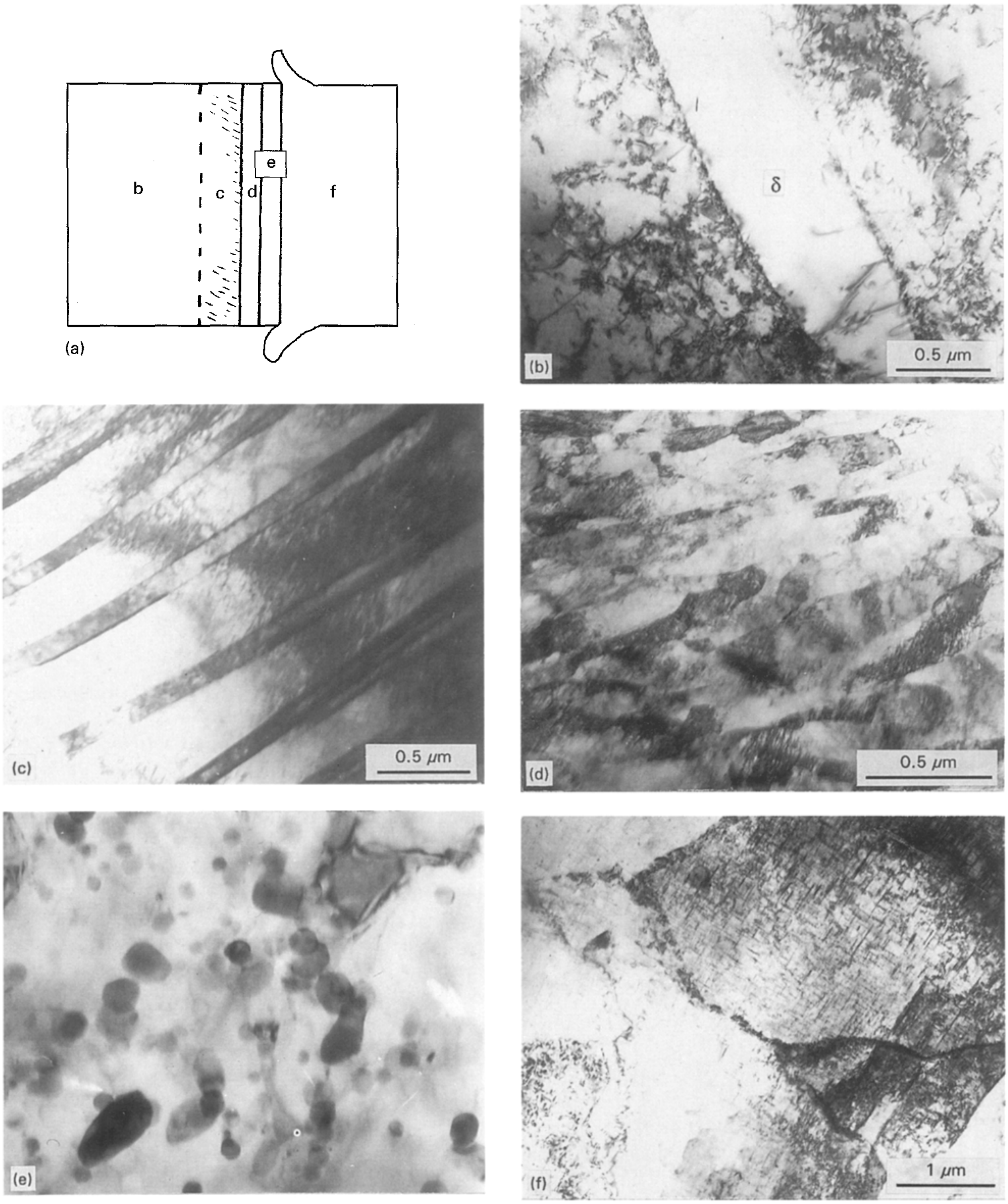


Figure 2 (a) Schematic drawing of the locations of the different microstructural features in dissimilar MMC/AISI 304 stainless steel joints. (b) As-received AISI 304 stainless steel microstructure showing delta ferrite in an austenite matrix. (c) Deformation twinning in the stainless steel substrate. (d) Fine-grained dislocation sub-structure of austenite. (e) $FeAl_3$ formed at the joint centreline region. (f) As-received MMC base material microstructure.

The high-temperature flow strength of AISI 304 stainless steel is much greater than in aluminium-based MMC base material. Consequently, almost all the axial deformation during the dissimilar friction joining operation occurs in the MMC substrate. This explains the planar shape of completed MMC/AISI 304 stainless steel friction joints (see Fig. 5). However, the dissimilar joint interface is only planar in macro-terms. The initial stage of the friction joining operation is characterized by local adhesion, seizure and shearing events that transfer material from one substrate to the other at the joint interface [8]. For example, Fig. 6a shows microcracking in a stainless steel base material immediately adjacent to the dissimilar joint interface.

The physical situation during the initial stage in friction joining has many similarities to that during sliding and rotational wear testing. Thus, the

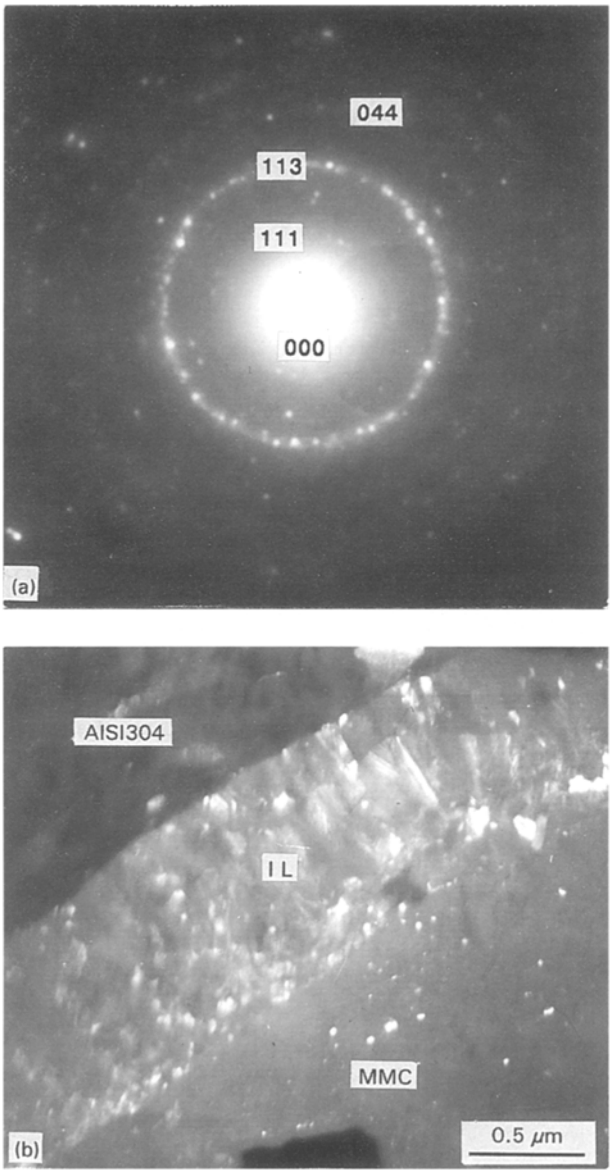


Figure 3 Electron diffraction analysis confirming the formation of Fe(Al, Cr)$_2$O$_4$ or FeO(Al, Cr)$_2$O$_3$ at the dissimilar joint interface (a) The selected-area diffraction pattern, (b) the interlayer (IL).

subsurface cracking shown in Fig. 6a is typical of that occurring during delamination wear of elastoplastic solids [9]. During the delamination wear process, crack formation is inhibited in the highly-deformed surface layer and strain accumulation at hard particles in the matrix (stainless steel) leads to crack formation and crack propagation parallel to the contact interface (this is the direction of maximum shear strain). This readily explains the delamination of stainless steel base material shown in Fig. 6b.

It has been suggested that preferential failure will occur in the lower strength (lower flow stress at high temperature) substrate during the initial stage in friction joining [3, 5, 7]. However, the results in the present study (e.g. Fig. 6b) and those found during sliding wear testing of MMC and steel base material combinations, do not support this view. For example, You *et al.* [10] examined the sliding wear behaviour of aluminium alloy (2124-AlCuMg) pins containing 20 vol% SiC particles in contact with 1045 carbon steel. The initial stage of sliding wear involved production of iron-based filings due to wear caused by SiC particles standing proud of the composite pin, i.e. SiC particles on the aluminium surface performed a micromachining operation of the stainless steel substrate and this resulted in a mixture of iron particles, SiC particles and aluminium-based material at the contact interface. In a similar manner, the results in the present study of MMC/AISI 304 stainless steel friction joints suggest that the stainless steel base material is severely deformed in a localized region immediately adjacent to the joint interface and that mechanical mixing of stainless steel and MMC base is an integral characteristic of the dissimilar friction joining operation.

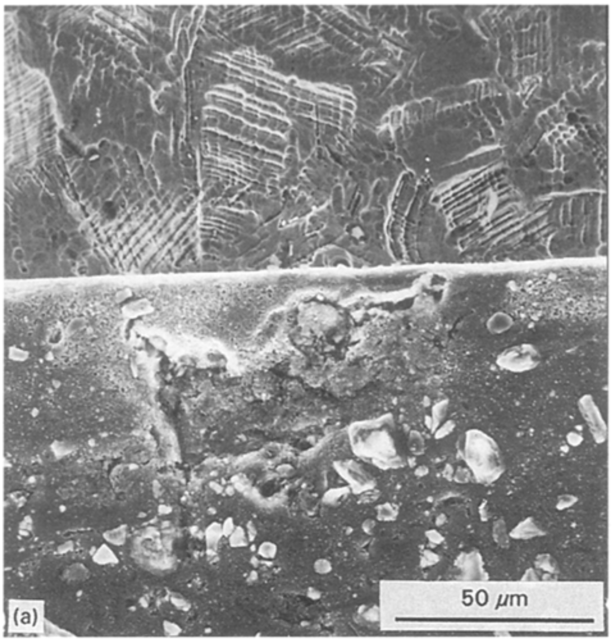
The stainless steel substrate is plastically deformed in the location close to the mid-radius in the dissimilar joint (see Fig. 4b). During the initial stage of friction joining (stage I), material at the periphery of the component heats up rapidly and yields. When this occurs, the joint region between the sample centreline and the periphery is heated preferentially. The heat input during friction joining depends on the axial pressure, radius and frictional coefficient, namely

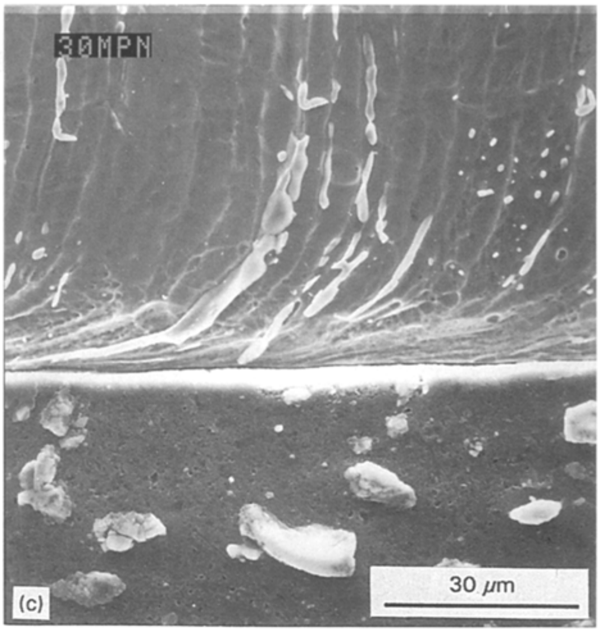
$$H = \frac{2\pi\mu N}{60} r\sigma_0 e^{2\mu(a-r)/h} \qquad (1)$$

where H is the heat input (W m^{-2}), μ the coefficient of friction, U the sliding velocity (m s^{-1}), P_a the axial pressure (MPa), N the rotational speed (r.p.m.), σ_0 the flow stress of the base material (MPa), h the depth from the contacting surface (m), r the radius (m), and a the distance $r + \delta r$ (m).

When a frictional coefficient of 0.3 is assumed during the MMC/AISI 304 stainless steel friction joining, the highest heat input occurs between the joint centreline and the joint periphery [11]. In this connection, a frictional coefficient value of 0.5 was assumed when modelling of the thermal cycle produced during AlMgSi and AlSiC friction joining [12] and a frictional coefficient value of 0.3–0.45 was found during sliding and rotational wear testing of MMC and stainless steel components [13].

The plastically deformed region observed in the stainless steel base material (see Fig. 4b) has its microstructural analogy in the MMC substrate. Discontinuous transition layers are preferentially formed in the MMC substrate in regions near the mid-radius of the dissimilar joint (see Fig. 7). Transition layers are regions of localized plastic flow and mechanical mixing in the MMC substrate [11]. In addition to transition layer formation, short-term testing (using a friction time 0.5 s) has confirmed that areas containing plastically deformed MMC base metal grains are preferentially formed close to the mid-radius locations in the dissimilar MMC/AISI 304 stainless steel joint (see Fig. 8). It is worth noting that the preferential formation of the regions containing partially recrystallized grains at the mid-radius location in the MMC substrate is only apparent in short-term friction tests. When the friction time exceeds 1.2 s, temperature equalization between the joint periphery and the centreline allows the region containing partially





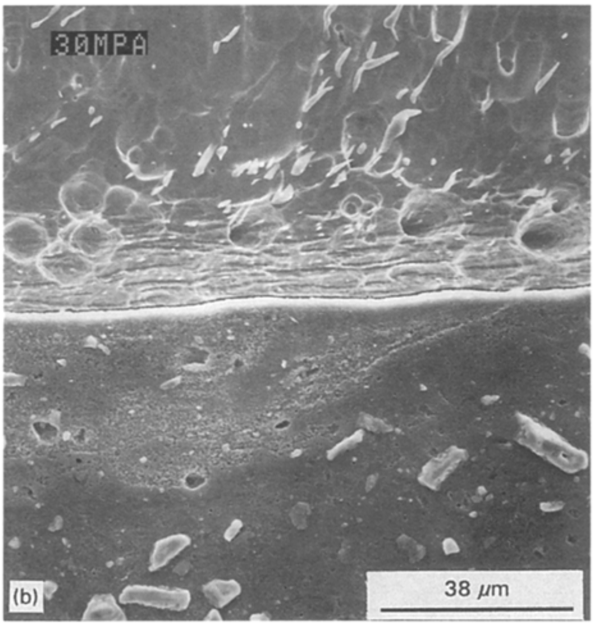


Figure 4 (a) Deformation twinning in the stainless steel substrate (in the region between the joint periphery and a location 4 mm from the joint periphery). (b) Plastically deformed region in the stainless steel substrate (in the region close to the mid-radius of the joint). (c) Microstructure at the joint centre line.

recrystallized grains in the MMC substrate to extend across the whole joint interface. However, the plastic flow-induced microstructural features observed in both the substrates (in Fig. 4b and 7) are clear evidence of the loading and heat-generation cycle applied very early in the friction joining operation.

3.2. Interlayer formation

The interlayer formed at the dissimilar joint interface comprised a mixture of polycrystalline Fe(Al, $Cr)_2O_4$ or FeO (Al, $Cr)_2O_3$ oxide and $FeAl_3$ (see Fig. 2e and 3). It has generally been assumed that intermetallic formation at the dissimilar joint interface determines the mechanical properties of aluminium/steel joints in particular, and dissimilar friction joints, in general [1–3, 14, 15]. However, in the present study, both oxide and intermetallic layers were observed at the dissimilar joint interface. Aluminium oxide is very stable and it is well known, for example, that the mechanical properties of diffusion-bonded aluminium alloys essentially depend on disruption and dispersion of this oxide film [16]. In contrast, oxides of metals such as titanium and copper are readily disrupted and are removed during the diffusion-bonding operation. In dissimilar friction joining, it would be expected that disruption and dispersion of aluminium oxide would be easily accomplished during the initial stage of the friction joining operation. However, the initial stage in friction joining may, in fact, provide the conditions for re-oxidation of the contacting surfaces. For example, oxidation at the contacting surfaces is an inherent feature during sliding and rotating wear testing of MMC and stainless steel components [13].

When an Fe(Al, $Cr)_2O_4$ or FeO(Al, $Cr)_2O_3$ oxide layer forms at the MMC/AISI 304 stainless steel joint interface, this will act as a barrier which will restrict intermetallic formation during the remainder of the

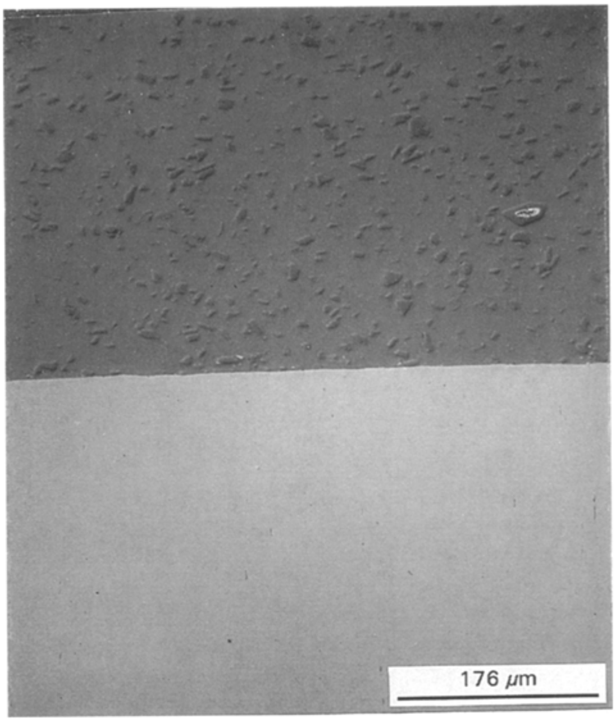


Figure 5 Planar joint interface formed in MMC/AISI 304 stainless steel joints.

friction joining operation. For example, Bedford and Boustead [17] have confirmed that the presence of $FeAl_2O_4$ spinel at the iron/aluminium interface severely retarded intermetallic phase formation in aluminium/steel bonds held at high temperature for long periods. In the present study, thick oxide layers and thin intermetallic films were observed at the joint centreline and at the half-radius location. In contrast, thin oxide layers and thick intermetallic films were observed at the joint periphery. These observations are consistent with the suggestion that oxide films restrict diffusion and growth of thick intermetallic films.

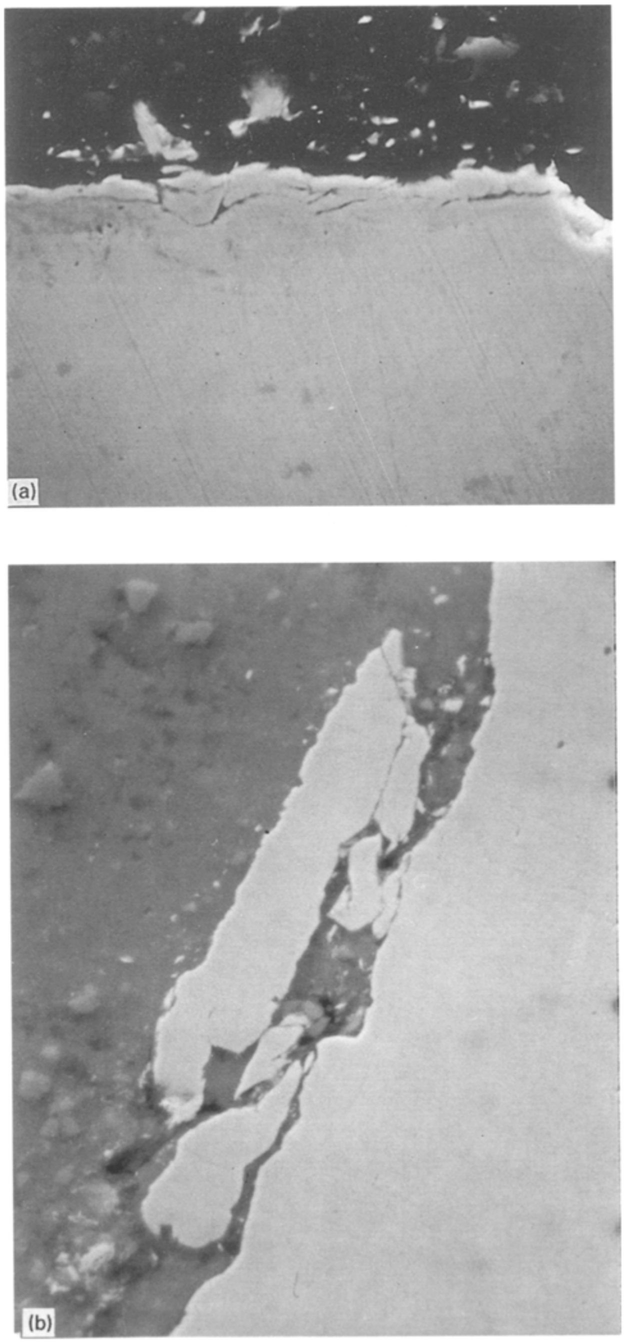
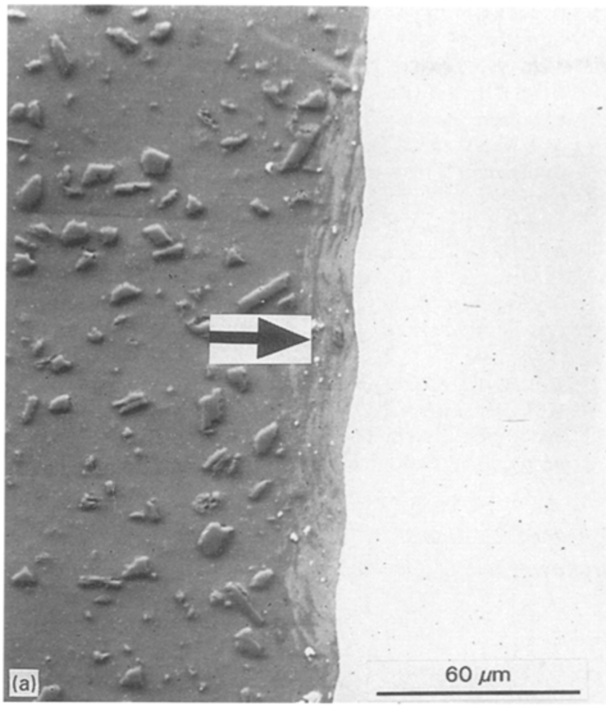


Figure 6 (a) Subsurface cracking in stainless steel base material at the dissimilar joint interface, $\times 2000$. (b) Delamination of stainless steel base material at the dissimilar joint interface, $\times 1300$.

3.3. Effect of rotational speed on joint notch tensile strength

It has been previously suggested that the use of high rotational speeds increases the joint interface temperature, promotes formation of thick intermetallic films at the dissimilar joint interface and results in poor joint mechanical properties [1, 2]. However, the results in the present study have confirmed that both oxide and $FeAl_3$ intermetallic films are retained at the dissimilar joint interface (see Figs 2e and 3). Also, the notch tensile strength of dissimilar MMC/AISI 304 stainless steel joints increased when the rotation speed increased from 500 r.p.m. to 1000 r.p.m., and at higher rotation speeds, there was no effect on notch tensile strength properties (see Fig. 9).

The results in the present study are therefore quite different from those indicated previously [1, 2]. They do not support the idea that enhanced interdiffusion, formation of thick intermetallic films and poor joint mechanical properties, are promoted when high rotation speeds are employed during MMC/AISI 304 stainless steel friction joining. It has already been confirmed that the optimum tensile strength of aluminium alloy A5056-H32 joints and MMC/MMC friction joints occurs when the highest frictional energy input per unit volume is applied [18, 19]. In a similar manner, Yashan *et al.* [5] have indicated that the tensile strength of inertia friction joints between aluminium and stainless steel attained a maximum value at critical given energy input ($J\ mm^{-2}$). With this in mind, it is therefore suggested that the low strength of MMC/AISI 304 stainless steel joints produced using a rotation speed of 500 r.p.m. is the result of insufficient frictional energy input during the dissimilar joining operation.



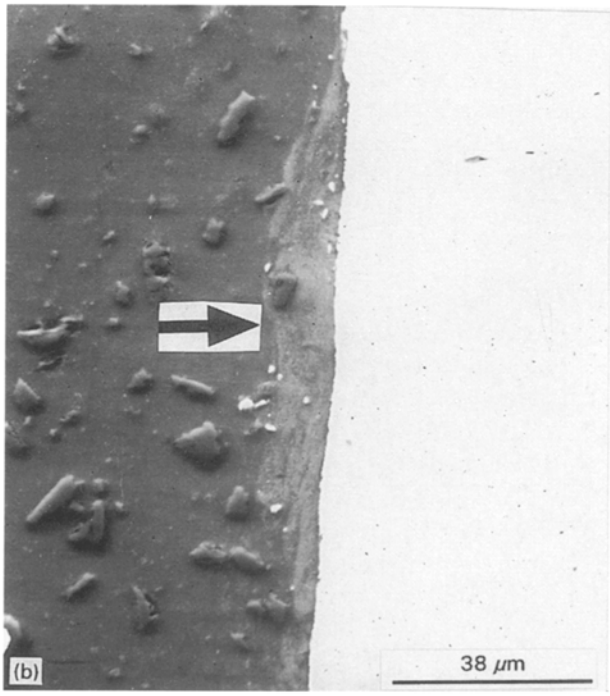


Figure 7 Transition layers formed in the MMC base material (in the region close to the mid-radius of the dissimilar joint). For a rotation speed of (a) 1000 r.p.m. and (b) 2000 r.p.m. The other joining parameters are indicated in Table I.

4. Conclusions

The microstructure and mechanical properties of dissimilar friction joints between aluminium-based MMC and AISI 304 stainless steel base materials were investigated. The following points have been confirmed.

1. The microstructural changes which occur in the stainless steel substrate comprise deformation twinning, formation of a fine dislocation sub-structure in austenite and plastic deformation. The stainless steel substrate was plastically deformed in the region close

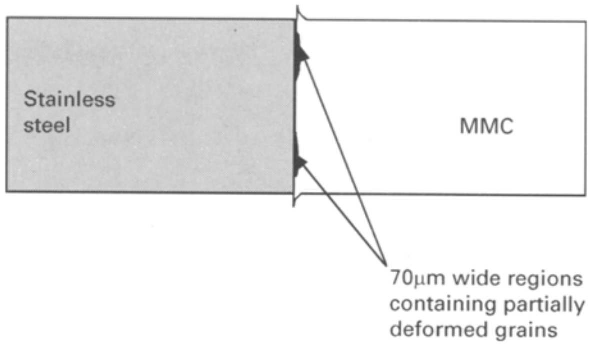


Figure 8 Schematic drawing showing the preferential formation of areas containing plastically deformed grains in the MMC substrate in a short-term test produced using a friction pressure of 30 MPa, a friction time of 0.5 s, a rotation speed of 1500 r.p.m. and a forging pressure of 30 MPa. After [11].

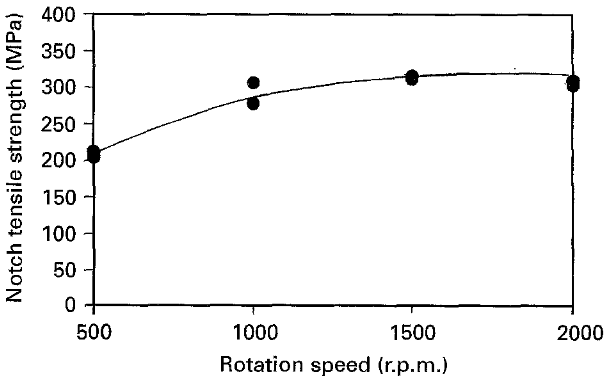


Figure 9 Effect of rotational speed on notch tensile strength.

to the mid-radius of the dissimilar joint. In a similar manner, 5–10 μm thick transition layers, comprising regions of locally plasticized MMC base material, were formed close to the mid-radius in dissimilar joints.

2. The interlayer formed at the dissimilar joint interface comprised a mixture of oxide (Fe(Al, $Cr)_2O_4$ or FeO·(Al, $Cr)_2O_3$) and $FeAl_3$ intermetallic phases.

3. The notch tensile strength of dissimilar MMC/AISI 304 stainless steel joints increased when the rotational speed increased from 500 r.p.m. to 1000 r.p.m., and at higher rotation speeds, there was no effect on notch tensile strength properties.

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