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Mechanical Properties of Particulate MMC/AISI 304 Friction Joints

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The influence of joining parameters (rotational speed, frictional time and pressure) on the notched tensile strength of dissimilar MMC/AISI 304 stainless steel friction joints is investigated. Frictional pressure and rotational speed have a statistically-significant effect on notched tensile strength values. The highest notched tensile strength properties occur in joints produced using a high frictional pressure (120 MPa) since extremely thin transition regions are produced at the joint interface and the area fraction of joint interface fracture in broken notched tensile test specimens is negligible.

The average particle diameter and inter-particle spacing decrease markedly in the region immediately adjacent to the bondline. It is suggested that the decrease in particle dimensions results from the combined effects of non-uniform plastic straining during friction welding and due to thermal shock and impaction of alumina particles on the contacting surface of the stainless steel substrate.

KEY WORDS: dissimilar friction welds; MMC/AISI 304; factorial experimentation; mechanical properties; particulate characteristics.

1. Introduction

Although much has been published concerning friction welding of dissimilar materials, little work has been carried out on dissimilar joining of particulate-reinforced base material. Kreye and Reiner[1] examined dissimilar friction joining of a mechanically-alloyed aluminium alloy base material (Dispal) containing a fine dispersion of alumina and carbide particles. The tensile strengths of Dispal/Dispal joints and dissimilar Dispal/carbon steel and Dispal/AISI 316Ti stainless steel joints were similar (300 MPa). Although refinement of the particle distribution was observed in completed joints, no explanation for this feature was indicated.

Aritoshi *et al.*[2] compared the friction welding characteristics of OFC (oxygen-free) Cu/Al and Cu–70wt% W/Al joints. CuAl and $CuAl_2$ intermetallics were detected in the thick transition region observed at the joint interface of OFC Copper/Al joints. In contrast, CuAl and $CuAl_2$ intermetallics were only detected in an extremely thin transition region at the joint interface of Cu–70wt%W/Al joints. The width of the interdiffused region produced during friction welding markedly increased when the high temperature flow strength of the Cu–W composite material decreased. With this in mind, Aritoshi *et al.*[2] associated the higher tensile strength of Cu–70wt%W/Al joints with formation of a thin interdiffused region at the joint interface.

The mechanical properties of dissimilar joints between aluminium and steel are largely determined by the presence of intermetallic films at the joint interface region. For example, FeAl and Fe_2Al_5 intermetallics have been detected at the joint interface in aluminium/stainless steel joints following heat-treatment.[3] In addition, Kikuchi *et al.*[4] indicated that Fe_2Al_5 and Fe_4Al_{13} intermetallics were formed in aluminium/mild steel friction joints.

It has been suggested that the detrimental influence of intermetallic films on joint mechanical properties becomes apparent when a critical transition (diffusion) layer thickness is exceeded in dissimilar joints.[1,5–7] For example, a critical transition layer thickness of between 1 and 2 μm has been observed in Al/Cu joints, in Al/carbon steel and Al/AISI 316Ti stainless steel joints,[1,5] in AISI 304L stainless steel/Ti joints,[6] in Nb/Armco iron joints[7] and during pressure welding of dissimilar materials.[3] It has been suggested that the detrimental influence of intermetallic particles in thick transition layers depends on the generation of triaxial stress due to the constraint imposed by the formation and clustering of needle-shaped intermetallic particles.[8]

Much research has been concerned with joining parameter optimisation so that thin transition (intermetallic) layers are formed during the dissimilar friction welding operation. For example, Jessop[5] examined dissimilar friction welding of pure aluminium to stainless steel and found that the highest shear strength occurred when thin intermetallic films were formed at the joint interface. Thick intermetallic films were observed at the joint interface in welds produced using high rotational speeds. Other investigators have suggested that intermetallic phase formation can be minimised by achieving a particular balance between rotational velocity and the axial deformation applied during friction joining,[3] by limiting the friction time[1] and by restricting the temperature range at the contact zone to 360–450°C during aluminium/stainless steel friction welding.[9] The influence of joining parameter selection on the retention of thick intermetallic layers at the joint interface has been associated with increased time and temperature for interdiffusion[5,10] and with the difficulty in squeezing-out ductile intermetallics from the weld zone.[1]

The present paper reports the first phase of a detailed investigation of the metallurgical and mechanical properties of dissimilar friction joints between aluminium-based metal matrix composite (MMC) base material and AISI 304 stainless steel. The aluminium-based MMC base material contains a dispersion of Al_2O_3 particles. The relation between bondline strength and joining parameters is examined using a combination of factorial experimentation and single variable testing. The metallurgical features of dissimilar MMC/AISI 304 stainless steel joint regions are investigated.

2. Experimental Procedure

The test materials comprised 19 mm diameter bars of alloy $6061/Al_2O_3$ base material (W6A.10A-T6) containing 10 vol% of 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, 1.0 wt% Si, 1 wt% Mg, 0.2 wt% Cr, balance aluminium.

Friction welding was carried out using a continuous drive friction welding machine. The influence of joining parameters on bondline mechanical properties was evaluated on the basis of a replicated 2^3 factorial experimental design. The underlying basis for the factorial experimental approach (see **Tables 1** and **2**) has been discussed in detail elsewhere.[11] The key variables were rotational speed, friction pressure and friction time during factorial experimentation. The different levels employed in the experimental designs are indicated in Table 1. It is generally recommended that the forging (upset) pressure should be at least twice the frictional pressure for production of welded joints free of unbonded regions. For this reason the upset pressure was set at twice the frictional pressure value during factorial experimentation. All other parameters were maintained constant during testing.

Table 1. Variable levels in the factorial design.

Variables In physical units	Coded	Levels set for experiments: Low (−1)	High (1)	Base (0)	Interval
Rotating speed (rpm)	N	1 000	1 500	1 250	250
Frictional pressure (MPa)	P_1	30	60	45	15
Frictional time (sec)	t_1	3	6	4.5	1.5

Table 2. Design matrix and notched tensile test results produced during factorial experimentation.

Trial No.	Design matrix N	P_1	t_1	Notched tensile strength (MPa)	Average	Total loss in length (mm)
1	−1	−1	−1	271, 284	277	2.7
2	1	−1	−1	257, 296	276	2.5
3	−1	1	−1	248, 293	270	7.6
4	1	1	−1	300, 337	318	8.6
5	−1	−1	1	222, 230	225	6.5
6	1	−1	1	254, 272	263	6.5
7	−1	1	1	282, 338	299	8.3
8	1	1	1	332, 343	337	9.0

The influence of incremental changes in rotational speed and frictional pressure on bondline notched tensile strength properties was also evaluated by varying the rotational speed from 500 to 2 000 rpm (all other parameters being held constant) and by varying the frictional pressure from 30 to 120 MPa (all other parameters being held constant). Again it should be noted that the forging pressure was set at twice the frictional pressure value in each case.

It has already been pointed out that conventional tensile and torsion testing of friction welded joints can produce results that may not reflect the actual mechanical properties which exist at the joint interface.[5,6,9] For example, the strength of the joint interface region can exceed the strength of a softer substrate. In addition, the mechanical properties at the joint interface are strongly affected by rigid restraint. In this case, the mechanical properties of material immediately adjacent to the bondline markedly affect deformation and failure of material at the joint interface region.[12] In conventional tensile testing, rigid restraint can substantially increase the strength of a dissimilar joint containing a weak joint interface region. As an example, dissimilar Ti/AISI 304L stainless steel joints with a tensile strength of 460 MPa can have extremely poor bend testing properties.[6] With this in mind, notched tensile testing was used as the measure of bondline mechanical properties in the present study. The test specimen designs employed during notched tensile testing are shown in **Fig. 1**. The V-notched design was used during factorial experimentation. The U-notched design was employed when examining the effects of incremental changes in rotational speed and in frictional pressure on bondline strength properties. It has been documented that the use of special forging die designs can counteract formation of unbonded regions at the base of the ejected flash.[13] Special forging dies were not employed in the present study and the use of the notched tensile test specimen design had the major advantage of counteracting problems caused by formation of local unbonded regions in some dissimilar joints.

Dissimilar joints were examined using a combination of optical and scanning electron microscopy. The dimensions and inter-particle spacing of alumina particles in the joint region were evaluated using image analysis (by measuring the alumina particles contained in regions located at different distances from the joint interface). During image analysis, the magnification was 200 ×; each calculated result is the average of five fields, measured fields having dimensions 250 μm × 50 μm. The

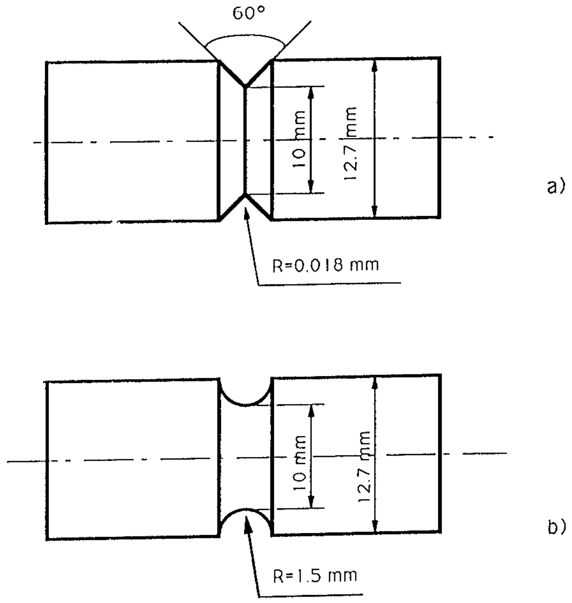

Fig. 1. Notched tensile test specimen configurations. (the V-notched design was employed during factorial experimentation; the U-notched design was employed during single variable testing)

Fig. 2. Microcracking in a thick transition region produced at the joint interface.

Fig. 3. Fracture surface morphology in a joint produced using $P_1=30$ MPa, $N=1\,000$ rpm and $t_1=1.5$ sec.

area fraction of bondline failure on broken notched tensile test specimens was measured by point-counting four 400× magnification SEM photomicrographs on each test section.

3. Results and Discussion

Notched Tensile Strength

The design matrix, factor levels and joint notched tensile strength properties are shown in Tables 1 and 2. In these tables, N is the rotation speed (rpm), P_1 is the frictional pressure (MPa) and t_1 is the frictional time (sec). Table 2 confirms that joint notched tensile strength properties are markedly affected by joining parameter selection and that there is a narrow operating envelope for attainment of optimum bondline notched tensile strength properties. In a similar manner, Murti and Sundresan[10] found that a restricted range of joining parameters produced optimum joint shear strength properties in dissimilar aluminium/austenitic stainless steel friction joints.

The notched tensile strength and joining parameter values are related as follows:

$$\sigma=283.3+13.9N+24.2P_1-2.4t_1 \quad \text{...........(1)}$$

where σ is notched tensile strength (MPa). The standard error of estimates during mechanical testing was 7.0 and consequently the statistically-significant variables are frictional pressure and rotational speed. In this connection, frictional pressure has a much larger influence than rotational speed on the notched tensile strength of dissimilar joints.

When the frictional pressure, friction time and rotational speed are set at their lowest levels, the completed joints contain microcracks in thick transition regions located at the joint interface (see **Fig. 2**). Residual stresses produced as a result of thermal expansion mismatch during cooling following completion of the joinining operation provide the driving force for microcrack formation at the joint interface. **Figure 3** shows that the fracture surface of broken notched tensile test specimens comprised regions of ductile failure and joint interface failure. Metallographic examination of cross-sections cut perpendicular to the fracture surface of notched tensile test samples confirmed that joint interface failure was associated with propagation of pre-existing microcracks and with preferential failure through thick transition layers located at the bondline.

The thickness of the transition layer varied from 2 to 11 μm at the joint interface of welds produced using a combination of low frictional pressure, friction time and rotational speed values. A discontinuous transition layer was observed with some regions at the joint interface exhibiting no observable transition layers (see **Fig. 4**). The formation of discontinuous transition layers of varying thickness at the joint interface readily explains

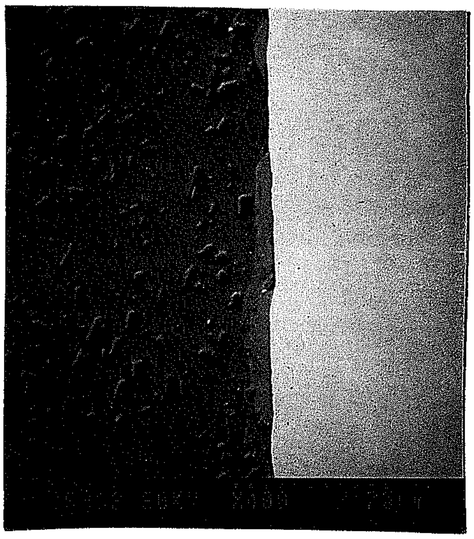

Fig. 4. Discontinuous transition layers located at the joint interface. (for $P_1=30\,\text{MPa}$, $N=1\,000\,\text{rpm}$ and $t_1=1.5\,\text{sec}$)

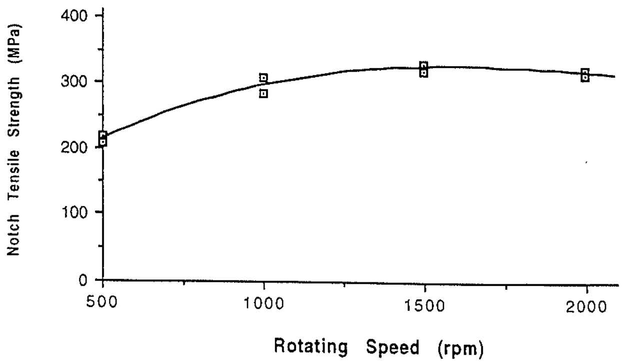

Fig. 5. Relation between rotational speed and the notched tensile strength. (for $P_1=60\,\text{MPa}$, $P_2=120\,\text{MPa}$, $t_1=4\,\text{sec}$)

the fracture surface morphology shown in Fig. 3, *i.e.*, localized regions of bondline failure separated by regions exhibiting ductile (microvoid coalescence) failure.

Factorial experimentation is particularly useful when determining the key parameters that affect bondline strength properties and the width operating envelope. However, single variable experimentation is more effective when a detailed analysis of joint metallurgical properties is required. The influence of rotational speed pressure on joint notched tensile strength properties is shown in **Fig. 5. Figure 6** shows the relation between frictional pressure and joint notched tensile strength. It should be noted that the forging (upset) pressure value was set at twice the friction pressure value during friction welding. It follows that the results shown in Fig. 6 are the produced by incremental changes in two parameters (friction pressure and forging pressure).

The notched tensile strength continuously increased when the frictional pressure increased. When the rotational speed increased, the notched tensile strength initially increased and then leveled out. Although the results in Figs. 5 and 6 are consistent with the data found during factorial experimentation, it is worth emphasising that the highest notched tensile strength properties were

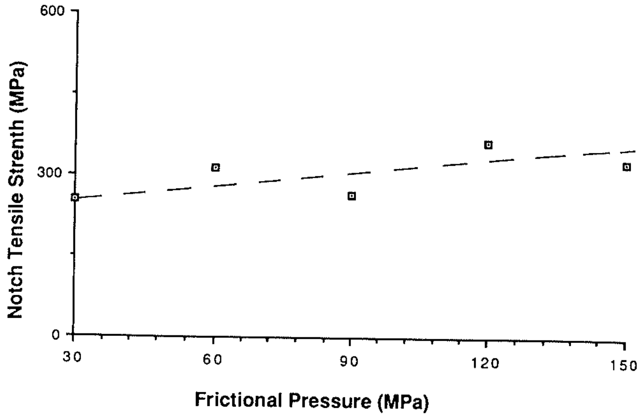

Fig. 6. Relation between frictional pressure and the notched tensile strength. (for $N=1\,500\,\text{rpm}$, $t_1=4\,\text{sec}$ and $P_2=2P_1$)

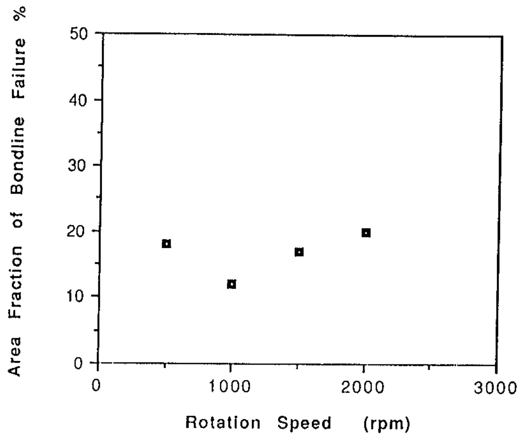

Fig. 7(a). Effect of rotational speed on the area fraction of bondline failure. (for $P_1=60\,\text{MPa}$, $P_2=120\,\text{MPa}$, $t_1=4\,\text{sec}$)

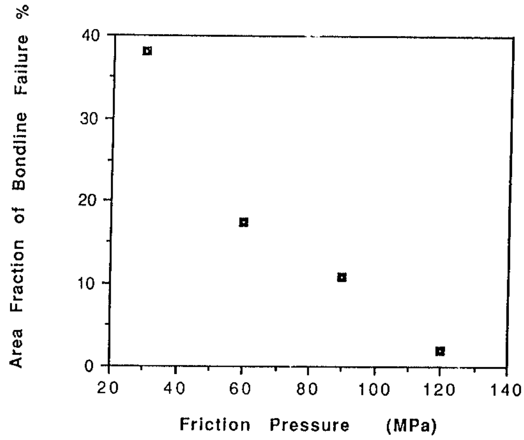

Fig. 7(b). Effect of frictional pressure on the area fraction of bondline failure. (for $N=1\,500\,\text{rpm}$, $t_1=4\,\text{sec}$ and $P_2=2P_1$)

produced using a high frictional pressure (120 MPa) during friction welding. Increasing rotational speed had negligible influence on the area fraction of joint interface failure in broken notched tensile test samples (see **Fig. 7(a)**). In contrast, increasing the frictional pressure markedly decreased the area fraction of bondline failure in broken notched tensile test samples and there was little evidence of bondline failure in dissimilar joints produced using a frictional pressure of 120 MPa (see **Fig. 7(b)**). In

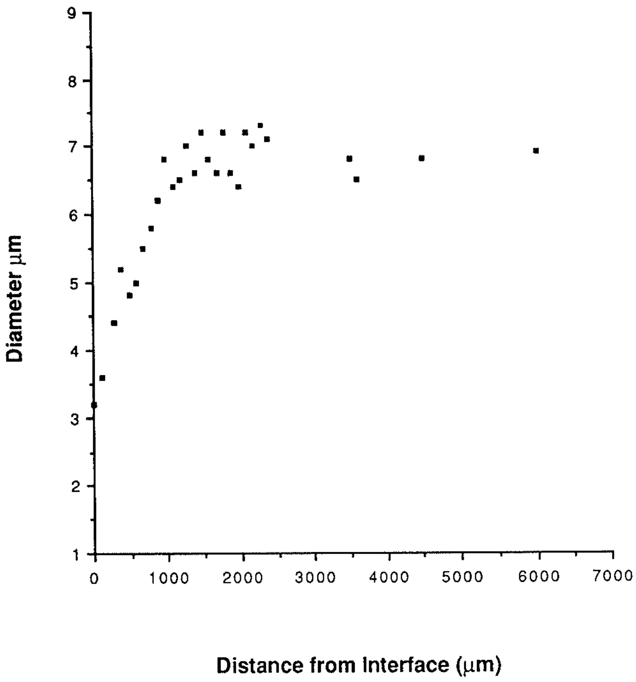

Fig. 8(a). The change in particle diameter at the joint region. (for $P_1=120$ MPa, $P_2=2P_1$, $N=1\,500$ rpm and $t_1=4$ sec)

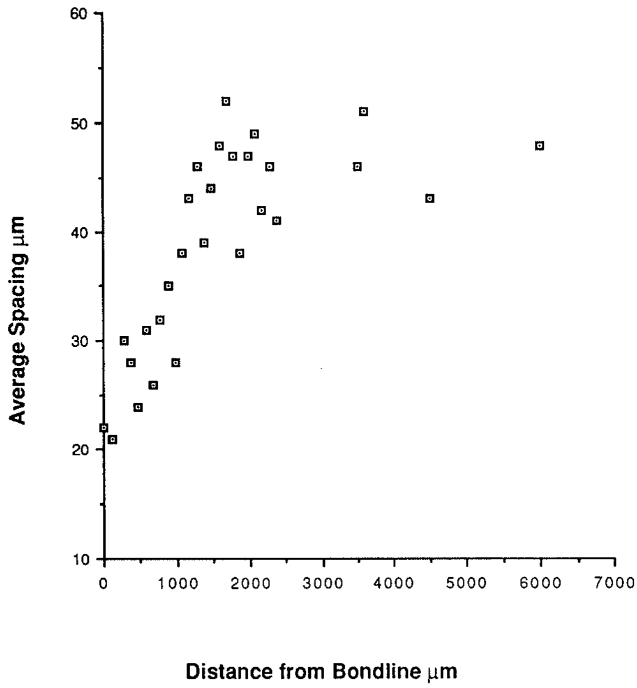

Fig. 8(b). The change in interparticle spacing at the joint region. (for $P_1=120$ MPa, $P_2=2P_1$, $N=1\,500$ rpm and $t_1=4$ sec)

addition, an observable transition layer was not found during SEM examination of this joint. These results readily explain the much higher notched tensile strength properties produced when this particular frictional pressure setting was used.

Particle Distribution in the Joint Region

Midling and Grong[14] noted that there was no evidence of particulate segregation at the joint interface in friction welded AlSiMg (A357) alloy base material containing 13 vol% of SiC particles. However, the particle size and

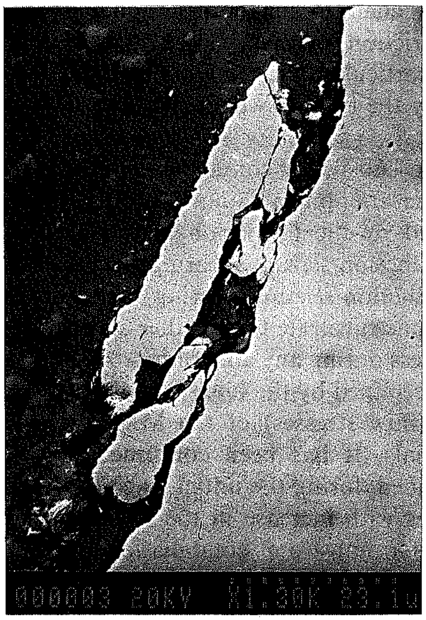

Fig. 9(a). Mixing of AISI 304 stainless steel and MMC base materials at the dissimilar joint interface.

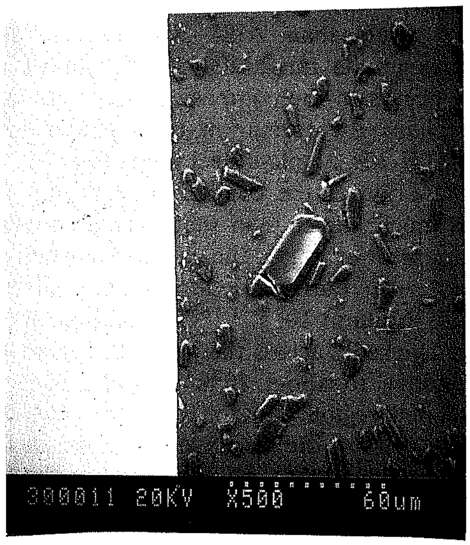

Fig. 9(b). Planar joint interface produced during dissimilar MMC/AISI 304 friction welding.

interparticle spacing in dissimilar MMC/AISI 304 stainless steel joints were significantly altered by the friction joining operation (see **Figs. 8** and **9**). The particle diameter and the interparticle spacing decreased in MMC base materials at the joint centerline, in the recrystallised region and in grains reoriented by upsetting during the joining operation. In addition, the average particle diameter and inter-particle spacing measured on the fracture surfaces of broken notched tensile test samples were much less than in the as-received MMC base material.

In dissimilar MMC/AISI 304 joints, the average particle diameter and inter-particle spacing decreased as a result of the friction pressure welding operation (see Fig. 8). It has been confirmed elsewhere that the particle diameter and spacing are markedly decreased when the frictional pressure and the forging pressure are increased. However, the particulate diameter and spacing are not affected by changes in the rotational speed during dissimilar friction welding.[15] An explanation for the changes in particulate diameter and inter-particle spacing produced by the friction welding operation is presented below.

When the substrates contact each other, the initial stage of the friction welding operation (Stage I) is characterised by the formation of a large number of localised adhesion/seizure/shearing events.[16,17] These localised adhesion/seizure/shearing events transfer material from one substrate to the other and *vice-versa*. The failure process during any adhesion/seizure and shearing event is determined by the high-temperature mechanical properties (flow stresses) of the MMC and AISI 304 base materials. It has been confirmed that the peak torque attained during the initial stage of friction welding increases with increase in the flow stress values of the dissimilar substrate combination.[5] Evidence of material transfer from AISI 304 stainless steel into the MMC base material is apparent in Fig. 9(a). When the steady-state period (Stage II) is attained during friction joining a fully-plasticised layer forms adjacent to the joint centerline. Since MMC base material has a much lower flow stress than AISI 304 stainless steel at temperatures in the range 400–500°C, almost all deformation occurs in the composite base material. This readily explains the formation of a planar interface in MMC/AISI 304 stainless steel joints (see Fig. 9(b)).

The strain rate at the contact zone during friction welding is much higher than that during conventional hot-working. For example, the calculated strain rate at the contact zone during friction welding of AlSiMg (A357) alloy is about $10^3\,s^{-1}$.[18] In addition, the strain rate varies markedly across the joint, *e.g.*, from $10^3\,s^{-1}$ at the bondline to about $10^2\,s^{-1}$ at the edge of the fully-plasticised region and to $10\,s^{-1}$ beyond that location. Also, during the steady-state period in friction welding, plasticised MMC base material in the contact region behaves like a rapidly-moving viscous fluid which impacts the planar stainless steel surface and is then continuously ejected from the joint. Following the steady-state period the joint region is upset and further material is ejected from the joint interface region. It follows that dissimilar friction welding involves severe non-uniform plastic deformation of the base material and impaction of particulate material on the contacting stainless steel surface. In addition, thermal stress produced by thermal expansion mismatch between the alumina particles and the aluminium matrix material may also contribute in the particle fracturing process. It is suggested that the changes in particle diameter and inter-particle spacing shown in Fig. 8 are produced by a combination of these effects.

The particulate fracturing process observed in dissimilar MMC/AISI 304 stainless steel friction welds leads to particle/matrix disbonding in regions close to the joint interface. The presence of regions where localized particle/matrix disbonding occurs provide preferential sites for failure during mechanical (notched tensile) testing. It follows that the mechanical properties of dissimilar MMC/AISI 304 stainless steel joints are determined by two critical effects, namely, by the formation of thick transition regions at the joint interface, and by the particle fracture and particle/matrix disbonding resulting from the dissimilar friction welding operation.

4. Conclusions

The influence of joining parameters (rotational speed, frictional time and pressure) on the notched tensile strength of dissimilar MMC/AISI 304 stainless steel friction joints was investigated. The principal conclusions comprise:

(1) Frictional pressure and rotational speed have a statistically-significant effect on notched tensile strength. Joints produced using a combination of low friction pressure, rotation speed and friction time values contained microcracks in thick transition layers located at the joint interface. Joints produced using a high frictional pressure (120 MPa) had the highest notched tensile strength properties since an extremely thin transition regions was produced at the joint interface and the area fraction of joint interface failure (on broken notched tensile test specimens) was negligible.

(2) The average particle diameter and inter-particle spacing decreased markedly in the region immediately adjacent to the bondline of dissimilar MMC/AISI 304 joints. The particle diameter and inter-particle spacing decreased as a result of the dissimilar friction welding operation. It is suggested that the change in particle diameter and in inter-particle spacing result from the non-uniform plastic strain applied during the friction welding operation and possibly as a result of impaction of alumina particles on the contacting surface of the stainless steel substrate.

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