Microstructure and mechanical behaviour of a SiC particles reinforced Al–5Cu composite under dynamic loading

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Abstract

The mechanical behaviour of a composite of Al–5Cu matrix reinforced with 15% SiC particles was studied at different strain rates from 1×10^{-3} to $2.5 \times 10^3 \text{ s}^{-1}$ using both a conventional universal testing machine (for low strain-rate tests) and a split Hopkinson bar (for tests at dynamic strain rates). Whilst the yield stress of the composite increases as the strain rate increases, the maximum flow stresses, 440 MPa for compression and 450 MPa for tension, are independent of strain rate. The microstructures and defect structures of the deformed composite were studied with both scanning electron microscopy and transmission electron microscopy and were correlated to the observed mechanical behaviour. Fracture surface studies of samples after dynamic tensile testing indicates that failure of the composite is controlled by ductile failure of the aluminium matrix by the nucleation, growth and coalescence of voids. © 1999 Elsevier Science S.A. All rights reserved.

Keywords: Aluminium composite; Silicon carbide particle; Dynamic loading; Microstructure; Hopkinson bar

1. Introduction

The advance and development of the aerospace industry has been giving a large driving force to the improvement of many classes of materials. The demand for better aluminium alloys is especially apparent, as they are used widely due to their high specific strength. However, the low specific modulus of aluminium alloys is a critical limitation of the monolithic alloys. This problem has been addressed by reinforcing the alloys with SiC particles or whiskers. Whilst there has been extensive study on the behaviour of SiC particles reinforced Al alloys upon quasi-static deformation [1–4], there are relatively few reports available on their dynamic response at high strain rates ($>10^3 \text{ s}^{-1}$) [5,6]. In order to use the composites more safely and efficiently, a more thorough understanding of their dynamic behaviour is clearly needed. In the present work, the dynamic mechanical response of a SiCp/Al–5Cu composite prepared by the powder metallurgical route was studied with a split Hopkinson bar in both tension and compression. The deformation microstructure and failure mechanism were studied by transmission electron microscopy (TEM) and scanning electron microscopy (SEM).

2. Experimental

2.1. Materials

The SiCp/Al–5Cu composite used here was fabricated by a powder metallurgy process using Al–5Cu powder of 280 grits and SiC particles with an average particle size of $3.5 \mu\text{m}$. Mixed powder with a SiCp volume fraction of 15% was compacted by hot compression under vacuum. The compacted billet was then extruded at 450°C to rods of 1.2 cm diameter with an extrusion ratio of 40:1. The general microstructure of the as-prepared composite is presented in Fig. 1, which indicates that the SiC particles are distributed evenly, and shows that there are no observable defects. Tensile samples of 3.2 mm diameter $\times$ 10 mm gauge length and compression samples of 8 mm diameter $\times$ 6 mm height were machined from the rods along the extrusion direction. Quasi-static compression and tensile tests were performed with an Instron universal-testing machine at a strain rate of 10^{-3} s^{-1} . Dynamic tensile and compression tests were carried out with a split Hopkinson bar at strain rates from $1 \times 10^3 \text{ s}^{-1}$ to $2.5 \times 10^3 \text{ s}^{-1}$ (for tension: $1 \times 10^3 \text{ s}^{-1}$ only). The working principle and set-up of a Hopkinson bar can be found in Ref. [7]. TEM specimens were made from thin slices (of about $150 \mu\text{m}$ thickness) cut parallel to the extru-

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Fig. 1. An optical micrograph of the as-prepared composite before deformation.

sion direction. The slices were ground mechanically to 70 μm thickness and then dimpled to 25 μm and finally ion thinned at 4 kV with an incident angle of 11°. The TEM specimens were examined with either a Philips CM20 or a JEOL 2000EX operated at 200 kV, whilst the fracture surfaces of the tensile samples were observed with an SEM.

3. Results and discussions

3.1. Mechanical behaviour

Stress–strain curves of the samples compressed at different strain rates are shown in Fig. 2. The yield stress (0.2% proof stress) of the composite increased from 180 to 220 MPa as the strain rates increased from 10^{−3} to 2.5 × 10³ s^{−1}. For all of the strain rates used, strain hardening was observed only when the plastic strain was low (<4%). A common saturation maximum flow stress of 440 MPa was observed after about 4% strain, irrespective of the strain rate used. A similar hardening behaviour has been reported in a SiCw/2124 Al composite [8].

Fig. 3 shows the stress–strain curve of a sample tested in tension at a strain rate of 1 × 10³ s^{−1}. For comparison purposes, the corresponding result for the sample tested in compression at the same strain rate is also plotted. It

Fig. 2. Compressive stress–strain responses of the SiCp/Al–5Cu composite at various strain rates.

Fig. 3. Compressive and tensile stress–strain curves of the SiCp/Al–5Cu composite at a strain rate of 1 × 10³ s^{−1}.

can be seen that the composite shows a similar trend of strain hardening in both tension and compression. The flow stress in tension is slightly higher and has a maximum saturation value of 450 MPa. This difference in mechanical behaviour under tension and compression is due to the residual stress in the composite. The residual stress was generated during the production of the composite, where cooling from high temperature resulted in different dimensional contractions in the matrix and the SiC particles due to their difference in thermal expansion coefficient [9].

As the saturation of flow stress at a relatively low strain (4%) is not common in monolithic aluminium alloys, it is considered to be an effect caused by the SiC particles. As pointed out by Hong and Gray [10], the work-hardening rate in composite materials should involve several terms, $\Theta = \eta(\Theta_h - \Theta_r - \Theta_{ad})$, where Θ is the net hardening rate; Θ_h , Θ_r and Θ_{ad} are the hardening components due to dislocation interaction, recovery and damage, respectively; and η is the ratio of the strength at two different temperatures or strength rates. From the present TEM studies, no sign of recovery can be found. Hence, the saturation of flow stress is considered to result from damage in terms of particle–matrix debonding and particle cracking. This damage is at free surfaces where dislocations can escape and thus relieve the elastic stresses that are due to dislocation pile-ups. Hence, the observed saturation stresses seem to represent some sort of critical stresses at which large-scale debonding and/or particle cracking occur.

3.2. TEM observations

3.2.1. Dislocations in the aluminium matrix

The microstructures of the aluminium matrix before and after dynamic compression are shown in Fig. 4(a) and (b), respectively. There are some dislocations in the aluminium matrix before the dynamic compression. These dislocations are generated during the production process of the composite due to the difference of thermal expansion coefficient of the SiC particles and the matrix. The dislocation density increased considerably after the dynamic compression (Fig. 4(b)), this is consistent with the strain hardening observed from the stress–strain curves.

Fig. 4. TEM micrographs of the Al-5Cu matrix: (a) before; and (b) after dynamic testing (strain rate= $1 \times 10^3 \text{ s}^{-1}$).

Fig. 5. TEM micrographs of regions near to small SiC particles (the dark-contrast particles of about 20–100 nm size): (a) before; and (b) after dynamic compression (strain rate= $1 \times 10^3 \text{ s}^{-1}$).

3.2.2. Dislocations near small SiC particles

Dislocation structures in the matrix regions near small SiC particles (about 20–100 nm diameter) before and after dynamic compression are shown in Fig. 5(a) and (b), respectively. It can be seen that the dislocation density increases substantially after dynamic compression; especially in regions near SiC particles where stress concentration is high. In fact, dislocation radiation from SiC particles to the matrix can also be observed (indicated by an arrow in Fig. 5(b)).

3.2.3. Defect structures in and next to large SiC particles

Defect structures inside large SiC particles and in their Al matrix surroundings before and after dynamic compression are shown in Fig. 6(a) and (b), respectively. Dislocations and evidence of shear deformation in SiC particles can already be found before the dynamic compression. It is considered that these defects are generated mainly during the production process of the composite when it was heavily (40:1) extruded. Upon dynamic compression, profuse stacking faults were produced in the SiC particles (Fig. 6(b)). Shear deformation associated with the stacking fault also generated dislocations in the aluminium alloy (indicated by an arrow in Fig. 6(b)). Curving of the stacking-fault traces indicates that fracture of the SiC particle had already occurred.

3.3. Fracture surface

The fracture surface of the sample tested in tension at a strain rate of $1 \times 10^3 \text{ s}^{-1}$ is shown in Fig. 7. The fracture surface is dominated by dimples of size about 1–3 μm . Occasional fracture and cleavage of large SiC particles can also be observed. No trace of SiC particle debonding from the matrix can be found. Voids, caused by plastic deformation, in the aluminium matrix near to SiC particles can also be observed (indicated by an arrow in Fig. 7). The fracture surface of the sample tensile tested at low strain rate ($1 \times 10^{-3} \text{ s}^{-1}$) shows no observable difference. These observations suggest that the composite mainly failed by nucleation, growth and coalescence of voids in the aluminium matrix at both quasi-static and dynamic strain rates.

The failure mechanism in Al alloy matrix–ceramic reinforcement composite is controlled by many factors. The nature of the matrix–reinforcement interface region is of particular importance. This interfacial effect on the failure mechanism during quasi-static deformation has been discussed extensively [1–4,11]. When the interface of the matrix and the reinforcing particles has a low cohesive strength, failure usually starts by cracking along the interface and is generally referred to as debonding. When the interfacial cohesive strength is high, the failure mechanism

Fig. 6. TEM micrographs of large SiC particles and the surrounding Al matrix regions: (a) before; and (b) after dynamic compression (strain rate = $1 \times 10^3 \text{ s}^{-1}$).

will be controlled by the strength of the matrix alloys. For a strong matrix, the composite will fail by cleavage along the low index planes of the ceramic particles, whilst for a matrix of lower strength, the composite will fail by nucleation, growth and coalescence of voids in the aluminium alloy.

Fig. 7. Fracture surface of the SiCp/Al-5Cu composite after the dynamic tensile test.

There are relatively few reports on the failure mechanism of dynamically loaded SiCp/Al alloy composites. Failure by interface debonding has been observed by Vaidya et al. [5,6] in a composite prepared by spray forming. No debonding was observed in the present work, which suggests that the composite prepared by the powder metallurgy technique used here has a strong SiC–aluminium alloy interfacial strength. Furthermore, the observed dimples and voids in the matrix suggest that failure in the present composite is that of such a type of composite having a strong interface but a weak matrix.

4. Conclusions

For the present 15 vol% 3.5 μm SiCp reinforced Al-5Cu composite prepared by the powder metallurgy method, the mechanical responses, defect structures and fracture mechanisms during dynamic deformation were studied. It was observed that at all of strain rates (1×10^{-3} – $2.5 \times 10^3 \text{ s}^{-1}$) used in the present studies, as the strain increases the flow stress increases to a maximum saturation value (440 and 450 MPa for compression and tension, respectively) at relatively low strains. These saturation stresses correspond to the critical stresses at which surface- or volume-defects are generated. Dislocations would then escape into these defects, prohibiting further strain hardening. From SEM and TEM observations these defects are micro voids in the aluminium matrix and fracture surfaces in the SiC particles. The failure mechanism in the present composite upon dynamic deformation is mainly the nucleation, grow and coalescence of micro voids in the aluminium matrix.

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