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Fracture behaviour of Al–Zn–Mg/SiC_p composites

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Al–Zn–Mg alloys reinforced with different volume fractions of SiC particulates were prepared by a liquid-metallurgy technique. The mechanical properties in uniaxial tension and compression were evaluated, and fractographic observations were made on the fracture surfaces. The distribution of SiC was quite uniform in the extruded condition, and the mechanical-property data show that the composite properties were inferior to those of the control alloy; they essentially showed a decreasing trend with increasing volume fractions. These observations can be explained in terms of the particle distribution, the porosity and the interfacial characteristics.

1. Introduction

The driving force for the development of metal-matrix composites (MMCs) has been the promise of larger quantum jumps in achievable properties than are possible with the development of conventional alloys [1]. In particular, light-weight aluminium-alloy-matrix composites (AMCs) are receiving increasing attention because of their improved strength, modulus, wear resistance and high-temperature strength relative to the values possible in conventional aluminium alloys [1–3].

In high-modulus particulate-reinforced AMCs, the strength and stiffness improvements are lower than those in continuous-fibre-reinforced composites, but these materials are isotropic and they are amenable to conventional metal-forming operations. Cast particulate-reinforced composites show promise, particularly for the automobile sector, due to their inherent simplicity and low cost [4–6]. A modification of liquid-metallurgy techniques is spray forming or spray deposition, in which a greater amount of freedom may be exercised in selecting the size and nature of the dispersoid. The use of particles of SiC as the reinforcing phase is receiving a good deal of attention because of their low density, high strength and easy availability. Efforts are, therefore, being made to develop low cost Al/SiC_p composites for various applications, and most studies appear to be focused on the 2xxx and 6xxx series of aluminium alloys. Among the Al alloys, it is well-known that the 7xxx-series (Al–Zn–Mg) alloys develop the highest strength, and they are used to make certain aircraft components. AA 7075 is regarded as one of the standard alloys for developing new combinations, and several alloy systems have been developed by tighter processing control of impurities, using suitable thermomechanical treatments etc. [7]. By dispersing SiC_p, there is a possibility of achieving even higher specific properties. McDanels [2], in the evaluation of several composite systems with discontinuous reinforcement, characterized the tensile properties of a 7075 alloy with SiC particulates. Manoharan and Lewandowski [8], by employing different heat treatments, studied the fracture characteristics of Al–Zn–Mg/SiC_p in both the under-aged (UA) and the over-aged (OA) conditions. In another study [9], fatigue crack growth was studied with respect to the SiC particle size and their volume fraction in an MB78 alloy matrix.

The property levels that can be achieved in MMCs depend on, besides the matrix, the spatial distribution of the dispersoids, the interface characteristics and the level of gross defects, like the porosity, etc. To some extent, these factors can be controlled through improvements in the processing route. To develop these composites for any application and to design any new matrix alloy specifically for making composites, a good understanding of the structure–property relationships in these materials is essential.

In this investigation, the properties of an Al–Zn–Mg alloy with dispersed SiC particles are discussed.

2. Materials and experimental procedure

A DTD 5024 (Al–Zn–Mg) alloy and SiC particles with an average size of 40 μm were used in this investigation. The composition of the base alloy is given in Table I. Composites containing various volume fractions of SiC were prepared using a modified liquid-metallurgy route [10]. This technique essentially involves addition of the dispersoid particles into the vortex formed in a pool of mechanically stirred molten alloy at 750–760 °C. A classified SiC powder was heated in a separate furnace for 2 h at 750 °C before addition. After the reinforcement was incorporated into the melt, stirring was continued for a further 10 min, and it was then cast into permanent moulds at

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TABLE I Chemical composition of the matrix alloy

	Chemical composition (wt %)								
	Zn	Mg	Cu	Mn	Si	Fe	Cr	Ti	Al
As-received material	5.81	2.97	0.46	0.57	0.16	0.19	0.037	0.02	Balance
Control alloy (0 vol % SiC)	5.67	2.82	0.43	0.56	0.14	0.18	0.036	0.02	Balance

a temperature of 710 °C. The period involved in the preparation of the composite (starting from the addition of particles to the final pouring) was kept as short as possible (< 20 min) so as to prevent any reaction between the SiC and the liquid metal. A control alloy, without any addition, was also prepared under similar conditions for comparison purposes; its composition is also given in Table I.

Cast billets were given a homogenizing and stress-relieving treatment at 450 °C for 16 h; they were furnace cooled and then hot extruded at an extrusion ratio of 15.5:1 at 525 °C. Samples, cut from suitable portions of the extrudants, were polished using standard metallographic techniques for microstructural analysis. Density measurements were made by a standard water-displacement method.

Tensile and compression tests were carried out on samples machined parallel to the extrusion direction in model 8032 Instron universal testing machine at a constant cross-head speed of 4 mm min^{-1}. The standard heat treatment chosen, before testing, was a solution treatment of 90 min at 485 °C, followed by rapid quenching in cold water and artificial ageing at 135 °C for 16 h. This approximates the T6 condition for the matrix alloy composition. Fractography was carried out in a Jeol scanning electron microscope.

3. Results and discussion

3.1. Microstructure

Typical optical micrographs of the control alloy and the composites containing 18 vol % of SiC are shown in Figs 1 and 2. There was no severe agglomeration. However, the as-cast microstructure was characterized by relatively-low-particle-density areas and porosity (Fig. 1). After hot extrusion, the same material exhibited an alignment of SiC particles (Fig. 2c) in the longitudinal direction. The particle distribution was quite uniform (Fig. 2b), although there were areas with

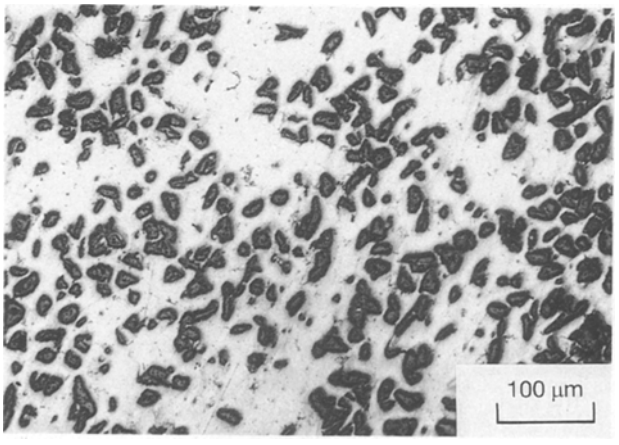


Figure 1 The as-cast microstructure of an 18.8 vol % composite, exhibiting particle-free areas and agglomerated regions.

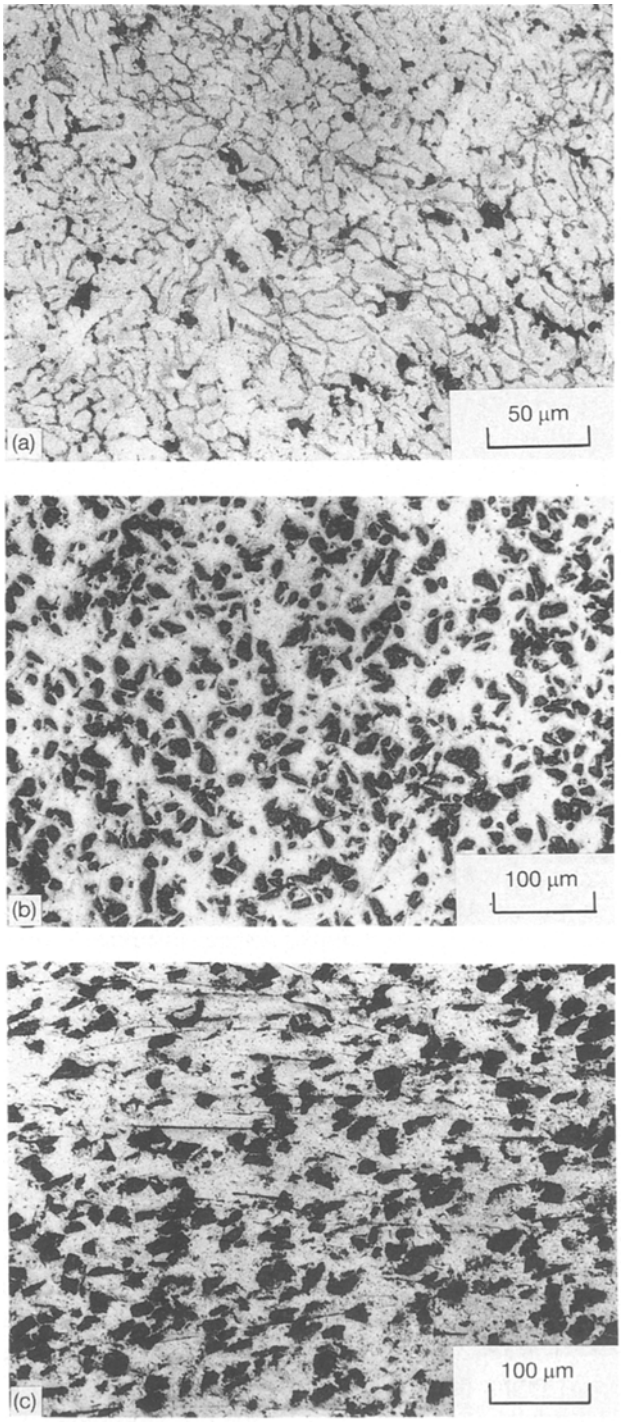
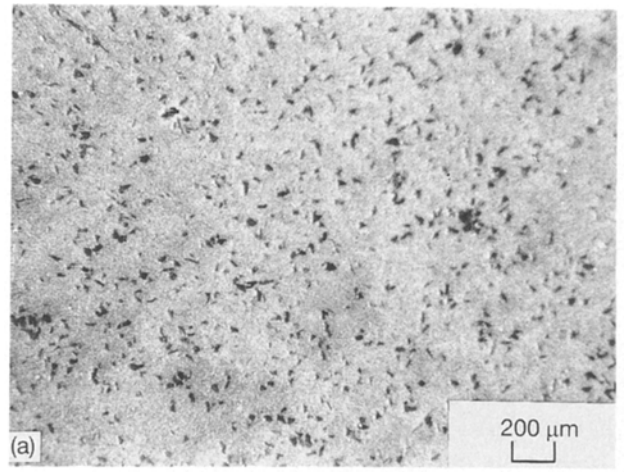


Figure 2 Extruded microstructures of: (a) the control alloy; (b) the 18.8 vol % composite, showing uniform distribution in the transverse direction; and (c) the alignment of particles in the extrusion direction.



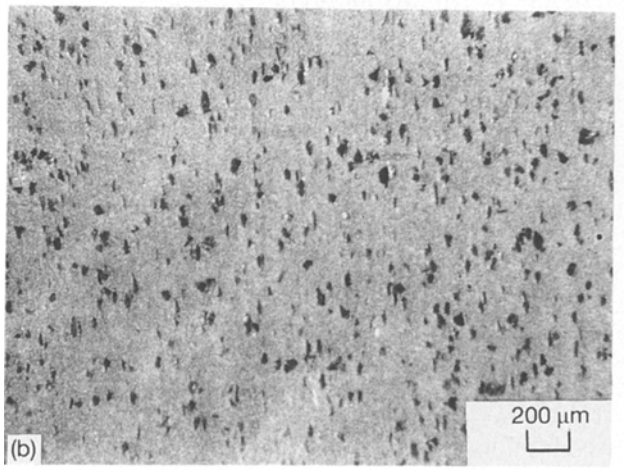
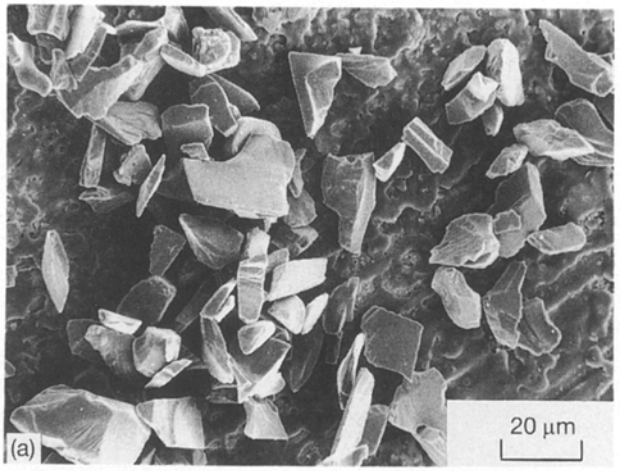


Figure 3 Low-magnification micrographs of the 8.8 vol % composite in (a) the transverse and (b) the longitudinal direction of extrusion.



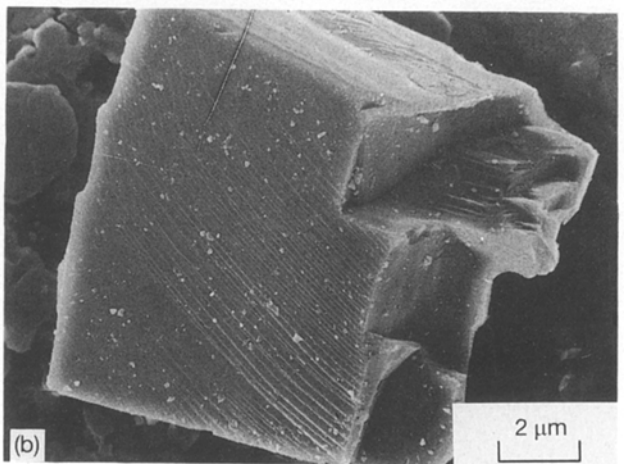


Figure 4 (a) As-received SiC particles with irregular shapes, and (b) a single particle showing regular lines on the surface.

a slightly larger number of density particles. The porosity was reduced to some extent in the extrudants. A low-magnification photomicrograph (Fig. 3) taken in a stereo microscope emphasizes these observations. The SiC particles used in this investigation were irregular in shape although some were cuboidal (Fig. 4).

The distribution of particles during processing and solidification is of interest, since the mechanical properties depend on the size, the size distribution and the orientation of the reinforcement. Similar to the observations in unreinforced alloys, the solidification characteristics influence the microstructure of the cast composites also. The distribution of particles is governed by the cooling rates encountered during solidification and by the settling velocities of the individual particles in the liquid melt. Depending on the ingot size, the spatial distribution of SiC changes to some extent from the chilled portion to the interior of the casting. SiC particles are pushed ahead by the moving solid/liquid interface or they become entrapped within the growing dendrites, and there is direct evidence [11] to show that decreasing the cooling rates results in greater clustering of the reinforcing particles. This problem is more pronounced for finer-sized particles. In the present case, because of the coarser size of the particles and the relatively small ingot sizes (75 mm diameter), the distribution of SiC_p, even in the as-cast condition, was relatively uniform.

Porosity is one common problem in Al-alloy castings, and in composites it takes on more serious dimensions. Fig. 5 shows the effect of the SiC volume fraction on porosity. Due to imperfect wetting and vortex creation during addition, air gets entrapped into the melt without being able to escape because of the increased viscosity of the melt due to particle addition. This also arises because of hydrogen evolution and shrinkage during solidification. But the fact that the porosity increases with the SiC_p volume fraction indicates that gases adsorbed on the particle surfaces may also be responsible. Improved and/or

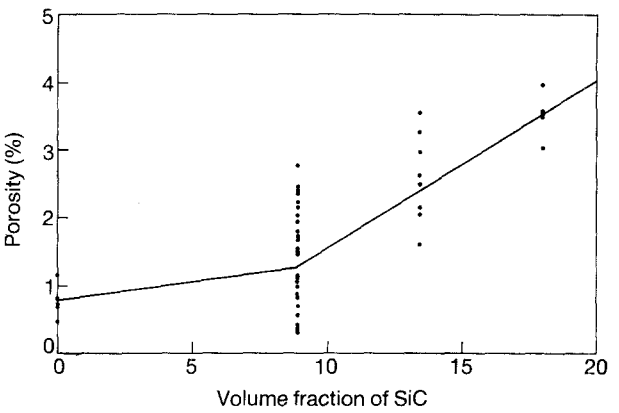


Figure 5 The porosity as a function of the SiC volume fraction.

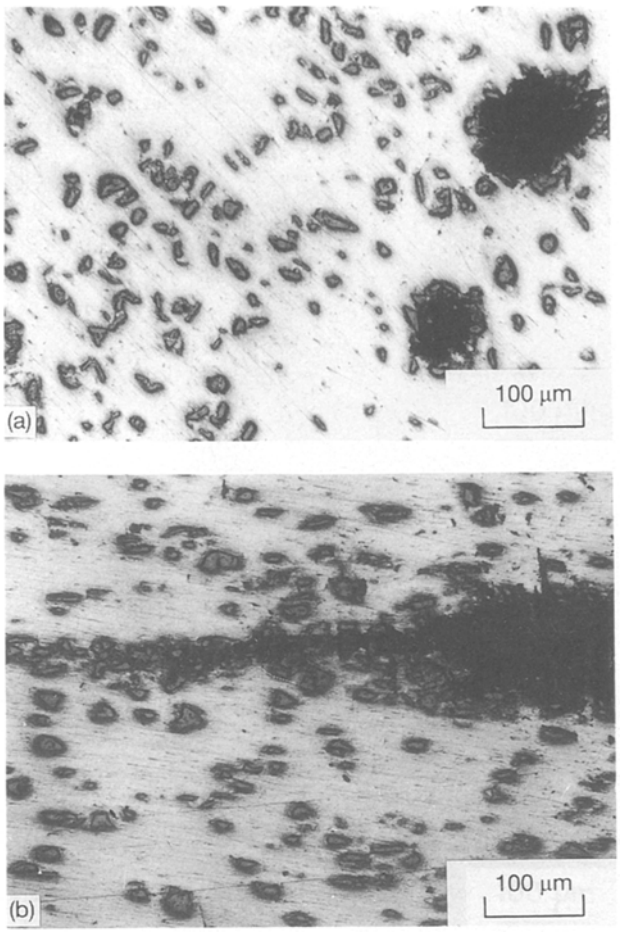


Figure 6 Optical micrographs showing particles clustering around a void: (a) in the transverse direction, and (b) in the longitudinal direction.

innovative casting methods may solve this problem to some extent.

Because of the particle flow in a compressed state during extrusion, the microstructure can become more homogeneous than that of the as-cast material. But, to some extent, any large inhomogeneity in the cast structure will be transferred to the extrudants. Fig. 6 shows the particles clustering around voids. The attrition during polishing of loosely bonded particles from the agglomerates which were not broken during extrusion gives rise to voids. Occasional particle breakage during extrusion was observed at the extrusion ratios used (Fig. 7).

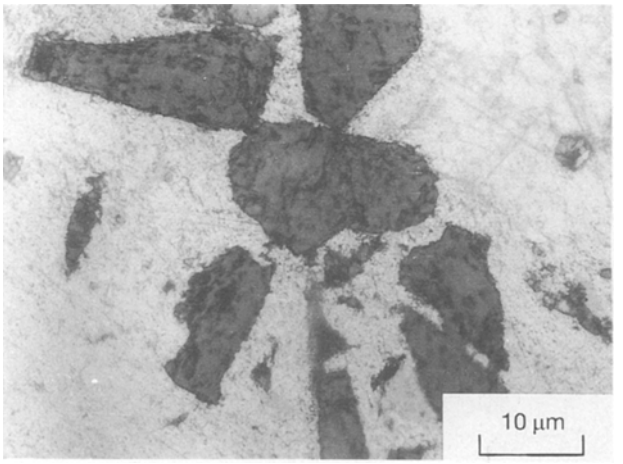


Figure 7 SiC-particle fracture after extrusion (for an extrusion ratio of 15.5:1).

The above results point to the fact that secondary processing can improve composite quality and lead to better distribution. In cases where the matrix is a wrought alloy, one can take advantage of this factor. In this study, hot extrusion, as a secondary processing operation, was found to improve the particle distribution and to decrease the porosity levels. Separate studies can be made to find the extent to which the particle distribution can be improved through other hot deformation processes. It is natural to expect that, as the volume fraction of SiC increases, the matrix response to deformation would be greatly influenced.

3.2. Mechanical properties and fractography

The mechanical-property data in uniaxial tension and compression are shown in Fig. 8. It is known that the ageing response of composites is reported to be different from that of the monolithic matrix alloy [12]; however, in this study the same heat treatment was maintained for all the materials. The elastic-modulus values increase with increasing volume fraction of the SiC (Fig. 4). With modulus values of 450 GPa for SiC and 71 GPa for aluminium, the value predicted by the rule of mixtures (ROM) is clearly an overestimate, as it is valid only for continuous-fibre reinforcement. The experimental points are close to the values predicted by the Tsai–Halpin equation [13]. The data indicate that, when compared with the matrix material, the ultimate tensile strength (UTS) values for tension and compression showed a decreasing trend with increasing particle loading. The strains at fracture for all composites are very low (about 1.8–2%) with respect to those for the control alloy (11%). But the respective

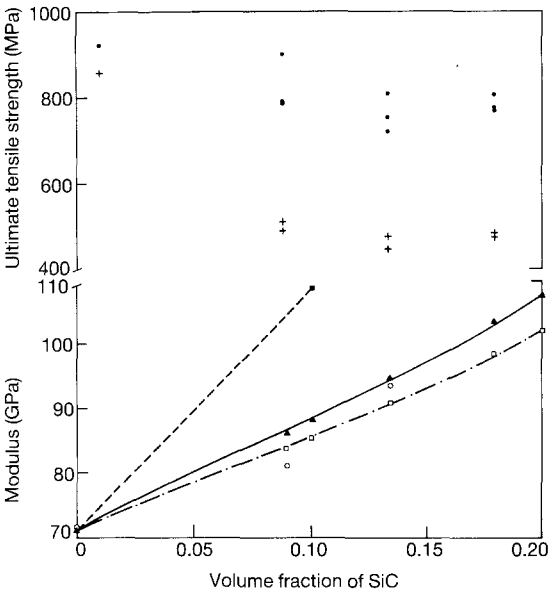


Figure 8 The effect of SiC on the ultimate tensile strength and elastic modulus: (●) compression, (+) tension, (■) rule of mixtures (ROM), (□) Tsai–Halpin equation [13], $s = 1$, (▲) Tsai–Halpin equation [13], $s = 1.5$, and (○) experimental values.

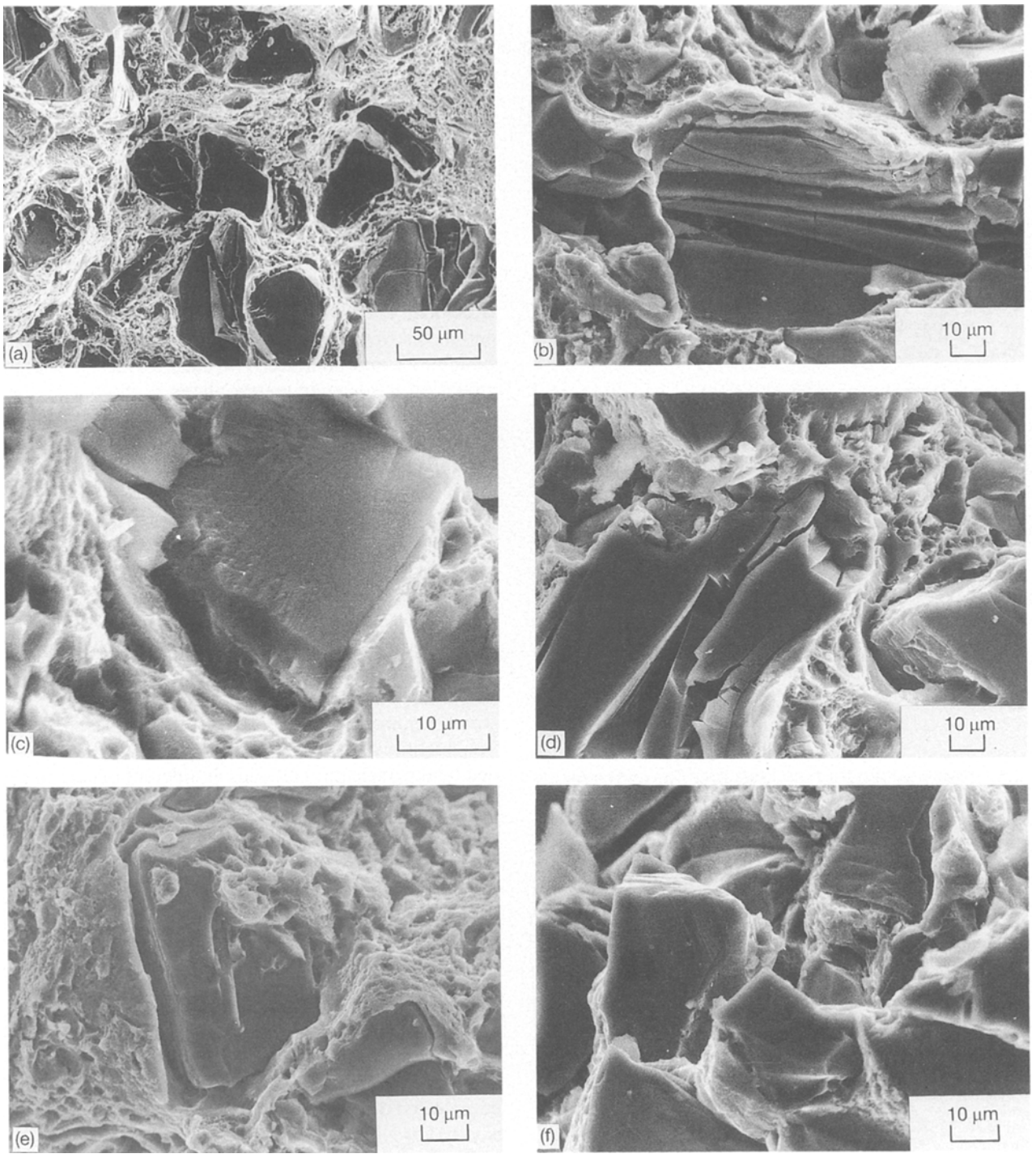


Figure 9 (a) General appearance of the tensile-fracture surface, (b) multiple particle cracking in a large particle, (c) poor interfacial wetting near one end of the particle, (d) particle pull-out and cracking, (e) clear separation of the interface on one end, and (f) fracture in a region with a high particle density.

strength values in compression for the composites are higher than those in tension. In particular, the fracture strains are an order of magnitude larger in compression (20–22%) than in tension.

Examination of tensile fractures observed by SEM (Fig. 9) revealed that the fracture is characterized by extensive particle fracture, particle cracking and particle pull-out. The relatively clean exposed surfaces of particles (Fig. 9a) point to the fact that they have cleaved during the fracture process. Large cracked particles (Fig. 9b) were observed, indicating that during the fracture process several particles which cracked could not participate in the deformation of the composite. Particle/matrix interface decohesion, as shown in Fig. 9d, was also seen in a few locations in the micrograph, indicating the interfacial weakness over the matrix. However, it is significant to note that decohesion is not observed everywhere nor is it observed all around a given particle (Figs 9c and e). Matrix adhesion to the particle was observed in several locations. In contrast, in the control alloy (0 vol % SiC_p), the fracture was characterized by dimples and the origin of the fracture was large intermetallic particles (Fig. 10) which form after solidification and during later heat treatment. It was observed that the greatly dimpled appearance of the

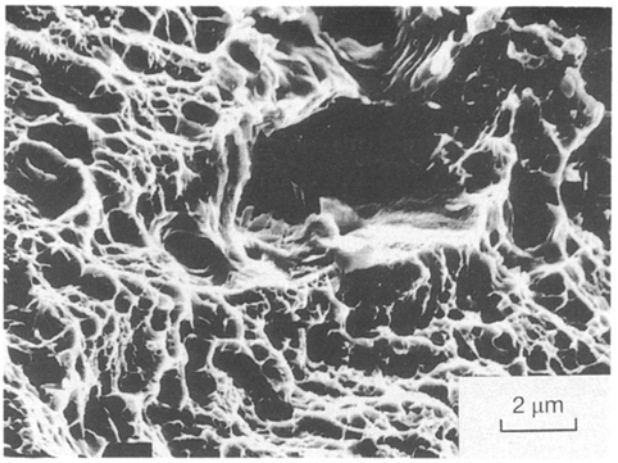


Figure 10 A tensile fractograph of the control alloy.

matrix portion of the composite was in contrast to the microcleavage-like appearance of the monolithic alloy.

Specimens tested in compression failed at an angle of 45° to the axis of loading. The fracture surface was shallow and flat and unlike that of tensile fracture (Fig. 11a). Examination at higher magnifications revealed that the areas around the particles exhibited certain deformation "damage". The matrix was smeared by the particle when being compressed (Fig. 11b). Particle fracture was not commonly observed. The fracture also exhibited features which were consistent with particles having moved considerable distances before final fracture.

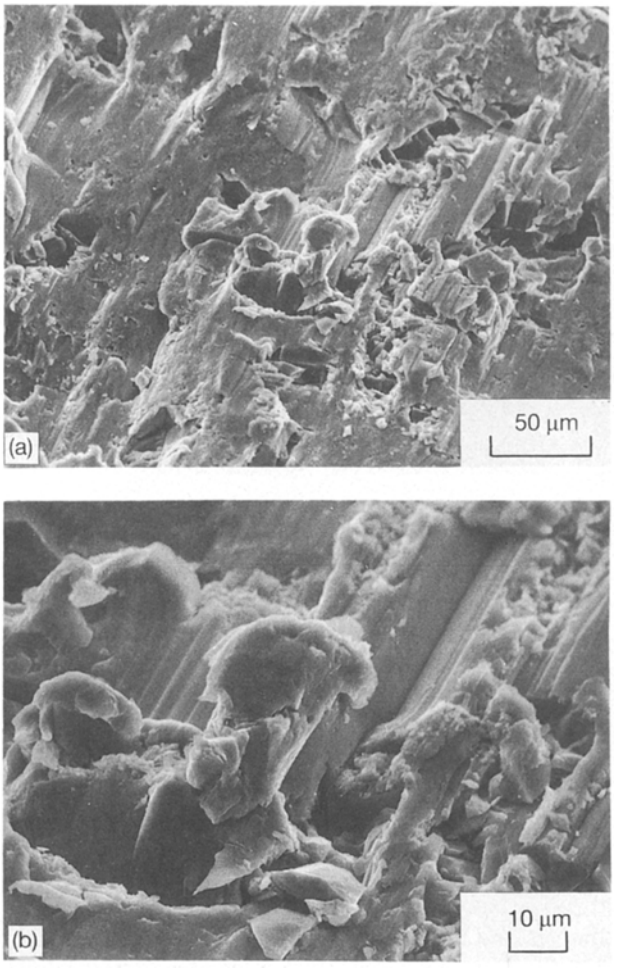


Figure 11 Compression fractographs showing: (a) a shallow-fracture surface, and (b) the matrix flow around the particle.

The general behaviour of MMCs is that there is an increase in stiffness and strength with a corresponding decrease in ductility and fracture toughness. In particular, the yield strength (YS) and UTS in Al/SiC_p systems containing a given volume fraction of reinforcement is dominated by the type of alloy used, since this determines, for a specific strengthening treatment, the matrix microstructure and any preferential distribution of alloying elements to grain boundaries and interface. The ageing kinetics in precipitation hardenable matrix alloys is affected by the difference in the coefficients of thermal expansion (CTEs) between the matrix and the dispersoids. Further, they are also dependent on the processing and forming techniques adopted which, as mentioned earlier, determine the particle distribution and the level of other defects (such as the porosity and inclusions). The interfacial strength, which depends on these factors, determines the effective load transfer. The higher levels of strength are attributed to different strengthening mechanisms, such as load transfer to the hard dispersoid phase, enhanced dislocation density due to differential CTEs of aluminium and SiC, and strengthening arising from constrained plastic flow. The ductility and strain-hardening coefficient are decided by the nucleation of voids at different places, since void growth and coalescence are extremely rapid. The sources for the initiation of voids are the large particles and coarse intermetallic inclusions in the matrix. Also, the porosity and associated agglomerates can cause premature fracture, and this can severely limit the strength to the extent that the matrix will not be able to participate in the deformation of the composites. In some cases, interface debonding at medium and large particles assists this process, although, normally, the strength of the interface is much higher in Al/SiC composites. Since there is a size distribution for the particles, naturally, the above possibilities can be expected during fracture, although the average size is very low.

Data on this particular system seem to be rather limited. In one study [2], it was found that the strength in a 7075/SiC_p composite increased relative to that of the respective control alloy. Shang and Ritchie [9] reported lower strength values for an MB78/SiC system than in the base alloy. But the reasons are not clear. As observed by Humphreys [14], high stresses will be developed in the reinforcement as the high-strength matrix is strained, and consequently particle fracture can occur, thereby decreasing the load-carrying capability of the composite.

In the 7xxx series MMCs, the development of precipitates in the matrix is accelerated faster than in the monolithic alloy [15]. Differential scanning calorimetry (DSC) studies have shown that the presence of SiC significantly affects the solid-state transformation kinetics in this alloy MMC [16]. Interfacial enrichment of Zn, Mg and Cu was observed in both UA and OA

composites with changes in the precipitation characteristics. In the OA condition, interface over-ageing took place [15].

In an Al–4.5 Zn–2.5 Mg alloy reinforced with a boron filament coated with a thin layer of SiC, Mg_2Si formed from the impurity silicon at the interface [17]. Fractographic observations confirmed the interfacial cracking around the fibres. Manoharan and Lewandowski [8] observed that the low toughness values in the OA composite (in comparison with the UA condition) were related to the void formation near the interface and the consequent propagation of cracks through the interface.

In the present case, the observations can be rationalized with respect to the combined effects of the porosity, the heat treatment and the interfacial characteristics. There is considerable scatter in the strength values; this is, presumably, because of the clusters of particles, along with the non-uniform nature of the porosity. The general trend of reduced strength can, again, be correlated with the void fraction that is present in the composites. At the extrusion ratios employed, the composites still exhibited some porosity which acts as sources of stress concentration in addition to agglomerated regions, that is, in regions with the lowest interparticle spacing. And as the volume fraction of hard particles is increased, the initial decohesion events become dominant, and fracture occurs by damage accumulation prior to any significant macroscopic strain. From the data available in the literature, it was logical to presume that the heat treatment that was given to composites must have over-aged the matrix to different extents, depending on the volume fraction of SiC_p.

The decrease in the yield stress in compression in composites over the base alloy indicates that high stresses are developed in the matrix, and the particles do not share the load from the matrix. The discontinuous interface observed in the fracture surface is also responsible for the decrease in strength in both tension and compression. Further studies are required to see whether deleterious interfacial precipitation is responsible for the decrease in strength.

4. Conclusions

1. A uniform distribution of particles can be achieved in composites made by the liquid-metallurgy technique.

2. The yield strength in compression decreased with increasing volume fractions of SiC when compared with that of the base alloy. The UTS values also showed the same trend.

3. For a constant volume fraction of SiC_p the tensile strength is higher in compression than in tension. The values of the percentage elongation were an order of magnitude larger than in tension, and they were essentially independent of the volume fraction.

4. The reduction in strength is attributed to processing defects and to interfacial characteristics.

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References

1. A. P. DIVECHA, S. G. FISHMAN and S. D. KARMARKAR, *J. Metall.* **33** (1981) 12.
2. D. L. McDANELS, NASA technical paper 2302 (July 1984).
3. S. V. NAIR, J. K. TIEN and R. C. BATES, *Int. Met. Rev.* **30** (1985) 275.
4. J. H. VACCARI, *Amer. Mach.* (1991) 42.
5. P. K. ROHATGI, *J. Metall.* **43** (1991) 10.
6. W. H. HUNT Jr, C. R. COOK and R. R. SAWTELL, in International Congress and Exposition, Detroit, Michigan, Society for Automotive Engineers (SAE) Technical Paper Series 910834 (1981).
7. T. H. SANDERS Jr and J. T. STALEY, in "Fatigue and microstructure" edited by M. Meshii (American Society for Metals, Metals Park, OH, 1979) p. 467.
8. M. MANOHARAN and J. J. LEWANDOWSKI, *Acta Metall. Mater.* **38** (1990) 489.
9. JIAN KU SHANG and R. O. RITCHIE, *Acta Metall.* **37** (1989) 2267.
10. U. T. S. PILLAI and R. K. PANDEY, *Comp. Sci. Technol.* **40** (1991) 333.
11. D. J. LLOYD, H. LAGACE, A. McLEAD and F. L. MORRIS, *Mater. Sci. Eng.* **A 107** (1989) 73.
12. T. CHRISTMAN and S. SURESH, *Acta Metall.* **36** (1988) 1691.
13. H. LAGACE and D. J. LLOYD, *Can. Met. Quart.* **28** (1989) 145.
14. F. J. HUMPHREYS, in "Mechanical and physical behaviour of metallic and ceramic composites", edited by S. I. Anderson, H. Lilholt and O. B. Pedersen (Riso National Laboratory, Denmark, 1988) p. 51.
15. M. STRANGWOOD, C. A. HIPPSLEY and J. J. LEWANDOWSKI in "Low density, high temperature powder metallurgy alloys", TMS-AIME Fall meeting, Detroit, MI; (TMS-AIME, Warrandale, PA, 1990).
16. J. L. PETTY-GALLIS and R. D. GOOLSBY, *J. Mater. Sci.* **24** (1989) 1439.
17. V. D. SCOTT, R. L. TRUMPER and MING YOUNG *Comp. Sci. Tech.* **42** (1991) 251.

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