JOURNAL OF MATERIALS SCIENCE 23 (1988) 83-92

Thermomechanical processing of aluminium-based particulate composites

O. P. MODI, A. H. YEGNESWARAN, R. ASTHANA, P. K. ROHATGI*
Regional Research Laboratory, Council of Scientific and Industrial Research, Hoshangabad Road, Bhopal 462 026, India

Thermomechanical processing of aluminium-based particulate composites containing dispersions of various soft and hard particles such as graphite, zircon, glass, alumina and mica in aluminium alloy matrices, with a view to improving strength and ductility for structural applications, has been discussed. The existing literature on the subject has been critically reviewed and analysed, and broad guidelines for optimum thermomechanical processing have been presented. Considerable improvements in strength and ductility of these composite materials have been reported after rolling, forging and extrusion due to fragmentation/fibrization of particles together with the refinement of matrix microstructure, annihilation of defects such as porosity, and texture hardening, etc. The influence of process variables, and of volume fraction, size and morphology of the particles on the strengthening mechanisms, fracture toughness and work-hardening behaviour of worked composites has been discussed.

1. Introduction

Metal–matrix composite materials produced by solidification techniques [1] are emerging as a new class of inexpensive, tailor-made materials for a variety of engineering applications. A considerable amount of work has been carried out in the last few years in the area of preparation, and characterization of structure and properties of these materials. Various hard and soft refractory particles (including short fibres and microballoons) of graphite, mica, silicon carbide, zircon and glass in various shapes and sizes have been dispersed by solidification techniques in different alloy matrices, principally aluminium based, to synthesize cast metal–matrix particulate composites [1]. Dispersions of particles such as graphite, mica, shell char, glass, zircon and sand in aluminium alloys produce cast composites possessing superior gall resistance, antiseizing properties, resistance to adhesive or abrasive wear, damping capacity and machinability [2–6]. Improvements in these properties are however, accompanied by decreases in strength and ductility of these cast metal–matrix particulate composites as shown in Figs 1 and 2. With a view to extending their applications to structural components, these composite materials are being processed by thermomechanical processing and squeeze casting in order to improve their strength and ductility. The purpose of this paper is to briefly review the current status of our understanding of thermomechanical processing of cast metal–matrix particulate composites, where the particle size is generally greater than 10 μm. The factors responsible for improved strengthening are then analysed with the available data. The important roles played by processing variables, microstructure and texture are emphasized with a view to improving the development of composites. The normal dispersion-hardened materials in which the particle size varies between 0.1 and 2.0 μm, the interparticle distance is less than 5 μm and the volume fraction of the dispersoids is generally below 0.10, are not included in the group of cast metal–matrix particle composites reviewed here.

2. Strengthening by thermomechanical processing

Thermomechanical processing such as extrusion, rolling and forging of aluminium-, magnesium-, copper- and nickel-based alloys containing dispersions of graphite, mica, silicon carbide, alumina, anthracite and glass has been reported to lead to considerable improvements in the tensile strength, yield strength, ductility and hardness of these composite materials. Hot extrusion of several compocast aluminium alloys (6061, Al–4% Cu–0.75% Mg and Al–4% Cu) containing a variety of non-metallic particles (alumina, SiC, mica and anthracite) and discontinuous fibres (graphite, boron) was carried out by Sato and Mehrabian [7]. These authors extruded 3.8×10^{-2} m diameter, 5.0×10^{-2} m high composite ingots at 733 K at rates of 1.7×10^{-3} to 7.6×10^{-3} m sec^{-1}. Reductions in area during hot extrusion were in the range of 78% to 96%. Tensile test bars were machined from all the extruded composites. These were soaked at 788 K for 15 h and aged at 428 K for 14 h. All composites were reported to deform uniformly and the dispersed non-metallic particles were found to align themselves in the direction of extrusion. Tensile testing of extruded composite specimens showed strength levels comparable to that of the matrix alloy. Finer non-metallic particles or discontinuous fibres

*Present address: Department of Materials, University of Wisconsin, Milwaukee, Wisconsin, USA.

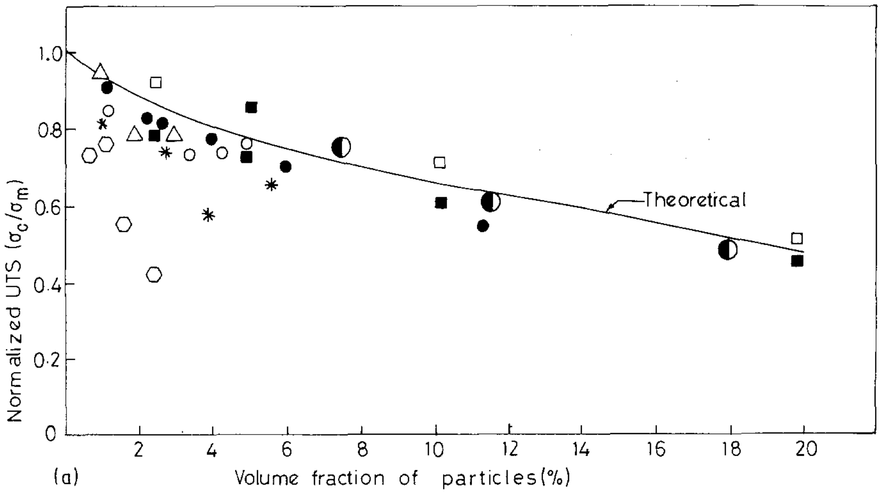
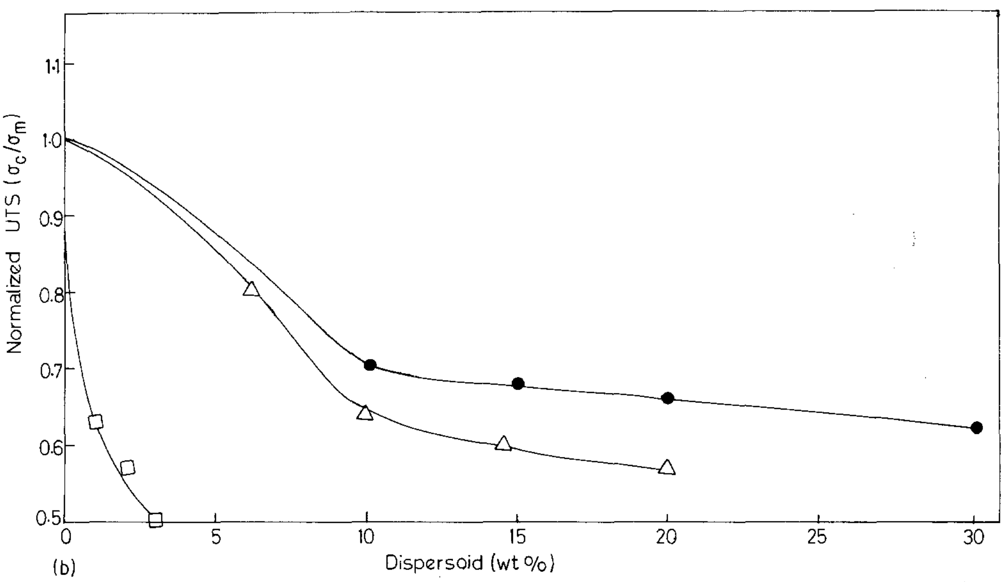


Figure 1 Normalized UTS plotted against (a) volume fraction of cast aluminium alloy composites containing graphite (● [38], ○ [39], △ [17], ■□ [40], * [41]) mica (⬡ [43]) and shell char (◐ [42]) particles [23], and (b) weight fraction of cast aluminium alloy composites containing hard particles of zircon (● Al–3% Mg–zircon [19]), sand (△ Al–11.5% Si–sand [37]) and alumina (□ Al–11.8% Si–alumina [31]).



dispersed in aluminium-alloy matrices led to improvements in strength and ductility. For example, addition of 15 wt %, 3 μm size alumina particles raised the yield and tensile strength of Al–4% Cu–0.75% Mg alloy from 227 to 302 MPa and 356 to 403 MPa, respectively [7]. The corresponding decrease in per cent elongation was from 25.8% to 12.5%. On the other hand, additions of 5 wt %, 5 μm size flakes of kaolin-mica were reported to improve the ductility and reduction in area, the latter from 14.5% to 52.6% at equivalent strength levels. Sato and Mehrabian [7] also reported that additions of relatively coarse mica (140–180 μm size) and anthracite (75 μm) particles did not severely impair the mechanical properties of hot-extruded composite samples. Copper-based alloys containing alumina particles in various amounts were found [8] to

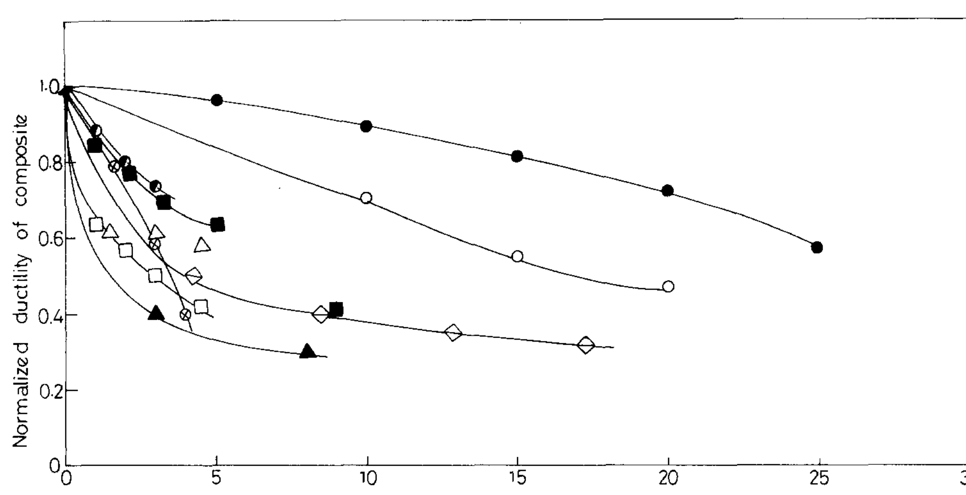


Figure 2 Normalized ductility of various cast aluminium-based particulate composites as a function of weight per cent dispersoid. (○) Al–3% Mg–zircon [19], (△) Al–3% Mg–graphite [19], (●) Al–3% Mg–zircon [32], (□) Al–11.8% Si–alumina [31], (◐) LM13–graphite [17], (◇) Al–graphite–Gradia [33], (■) Al–11.8% Si–1.5% Cu–graphite [34], (▲) Al–11.8% Si–3% Mg–shell char [35], (⊗) Al–11.8% Si–1% Mg–titania [36].

attain considerable increase in the high-temperature strength up to closely below the melting point of the matrix alloy as a result of extrusion.

Bergmann [9] has carried out extrusion of aluminium–glass particle composites at 723 K prepared by powder-metallurgy techniques. The glass particles were found to fibrize (3 to 4 μm diameter) in the extruded samples and the tensile strength increased four to five times the strength of the same composite before extrusion. Glass acts as a lubricant during extrusion and fibrization gives additional strength to the composite after extrusion. In another study on aluminium–glass particle composites synthesized by the powder metallurgy route, Badzoich *et al.* [10] obtained glass fibre-reinforced aluminium composites by extruding composites of metal powders and glass particles. Considerably lower extrusion pressures were required to deform the glass composites because of the lubricating action of glass during extrusion. Powder metallurgy techniques, however, resulted in high porosity levels and relatively poor corrosion resistance of the composite.

Keshavaram *et al.* [11] have prepared commercially pure aluminium–glass and aluminium–flyash particle composites by a liquid metallurgy technique. It was found that on adding surface active agents (magnesium) to molten metal during composite preparation, up to 8 wt % glass and 10 wt % flyash particles could be introduced in composite castings with ease, whereas a maximum of only 2.5 wt % particles could be introduced in the alloy in the absence of magnesium. Composite specimens of 15 cm diameter were extruded at 753 K and with an extrusion ratio of 12 : 1 in a 1250 T horizontal press. The extrusion loads were 1180, 930 and 1230 ton for Al–2% Mg, Al–2% Mg–8% glass, Al–2% Mg–10% flyash particle composites, respectively. The strength of the composite increased by over three times after extrusion, from 63.7 $MN\,m^{-2}$ in the as-cast condition to 190.12 $MN\,m^{-2}$ after extrusion in the case of the aluminium–glass composites. In the case of Al–2% Mg alloy containing 10 wt % flyash the UTS increased from 56.84 MPa in the as-cast condition to 182.28 MPa after extrusion. Part of the improvement in strength of the composite is, however, attributable to the matrix itself, because in the absence of the particles, the strength of the extruded Al–2% Mg alloy itself increased two-fold as compared to the as-cast condition.

In a manner similar to the fibrization of glass, graphite particles were also reported [12] to deform into stringers as a result of hot extrusion of Al–11.8% Si–3 wt % graphite particle composites. The cast ingots were soaked for 90 min at 723 K and extruded at 723 K at a rate of 1.2×10^{-1} m sec^{-1} into rods of 6×10^{-3} m diameter. A 95% reduction in area was achieved by extrusion. The UTS of the composite increased from 130 $MN\,m^{-2}$ in the as-cast condition to 208 $MN\,m^{-2}$ after extrusion. The effect of working on wear properties of these aluminium–graphite composites also has been characterized because these composite materials are principally meant for tribological applications. Hot-worked (rolled and forged) Al–Si–graphite composite specimens showed lower wear loss compared to as-cast samples under identical conditions [13].

Modi *et al.* [14] carried out rolling and forging of LM13–3 wt % graphite particle composite at 723 K. The UTS of the composite, rolled and heat treated, increased from 90 MPa in the as-cast condition to 150 MPa after 35% reduction; in the forged and heat-treated condition, the UTS increased to 135 MPa after 33% reduction. These authors also reported deformation of graphite particles, initially homogeneously dispersed in the cast matrix (Fig. 3a), into stringers (Fig. 3b) along the working direction as a result of working. The aspect ratio of the stringers increased with increasing amount of deformation; higher aspect ratios were obtained along the long transverse direction of working as shown in Fig. 3b.

In another study on processing of aluminium–graphite composites, Modi *et al.* [15, 16] carried out thermomechanical processing of Al–1% Mg alloy containing up to 2.5 wt % graphite particles by hot extrusion, and rolling and forging. Graphite (60 to 150 μm size) containing alloys were prepared by a vortex technique [17]. One set of samples machined to 50 mm diameter was extruded at four different temperatures (623, 673, 723, 753 K) using extrusion ratios of 9.77 and 25. Another set of samples was rolled at 673 K to 20% reduction followed by forging. Reduction per pass during rolling was kept below 2% and after each pass the samples were annealed for 10 min

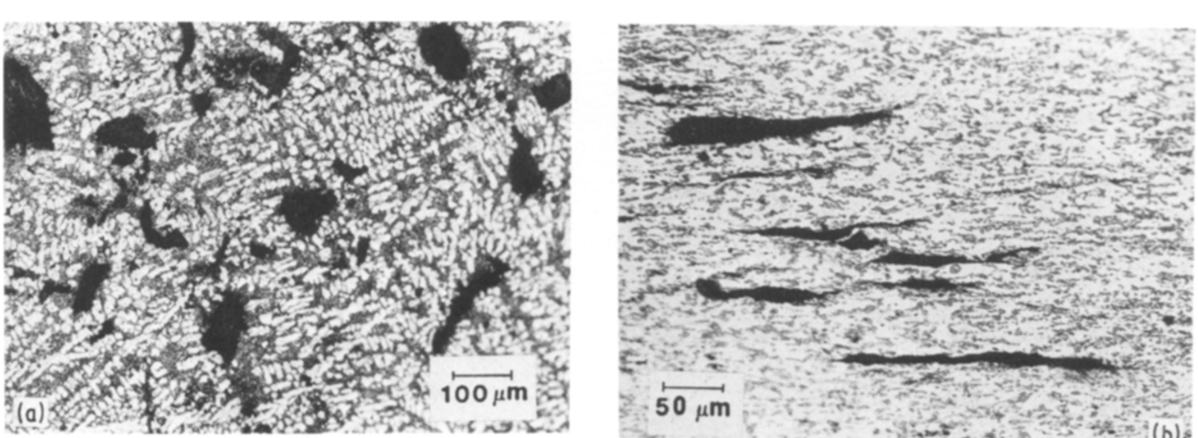


Figure 3 Micrographs of (a) cast and (b) rolled (72% reduction) LM13–3 wt % graphite particle composite.

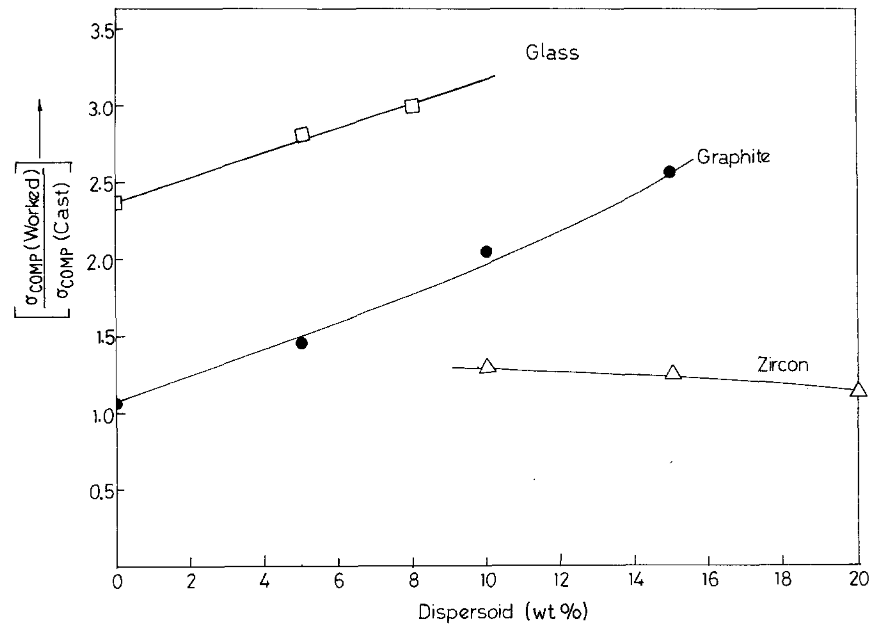


Figure 4 Normalized UTS of worked composites as a function of weight per cent dispersoid for aluminium alloy composites containing glass, graphite and zircon particles [11, 18, 19].

at 673 K. Care was taken to avoid edge cracking during rolling. Tension and hardness tests carried out on deformed composites generally confirmed earlier results showing improvements in strength and ductility.

Fig. 4 shows a plot of normalized UTS of worked composites (normalized with respect to the UTS of the corresponding cast composite) for aluminium alloys containing graphite [18], zircon [19] and glass [11] particles as a function of the weight per cent dispersoid. It can be seen from this figure that the relative effectiveness of thermomechanical processing improves with increasing amount of dispersoids for glass and graphite particle composites. On the other hand, zircon particle composites show a steady decrease in the value of normalized UTS with increasing weight per cent zircon in the composites. The increasing size of the particulate phase had a beneficial effect on the strength and ductility of Al–3% Mg–zircon and Al–3% Mg–graphite particle composites in the size range 100 to 200 μm. Fig. 5 shows variation of normalized UTS and percentage elongation of extruded aluminium alloy graphite composites as a function of the extrusion temperature. Data from two studies [16, 18] are shown in the figure. The normalized UTS (Curves 3 and 4) of extruded composites decreases with rise in extrusion temperature due to increased plasticity of the alloy matrix. The normalized per cent elongation shows an increase with an increase in the extrusion temperature in the data of Modi *et al.* [16] (Curve 1), whereas it shows a decrease (Curve 2) for Yuasa and Morooka's data [18]. In the latter study, aluminium–graphite composites containing nickel-coated graphite particles were used. Conceivably the dissolution of nickel coating from the particle surface at high temperatures during extrusion, as well as during preparation of the composite, could have caused solid solution strengthening of the aluminium alloy matrix resulting in a decrease in the ductility. In another study of hot deformation of aluminium alloy

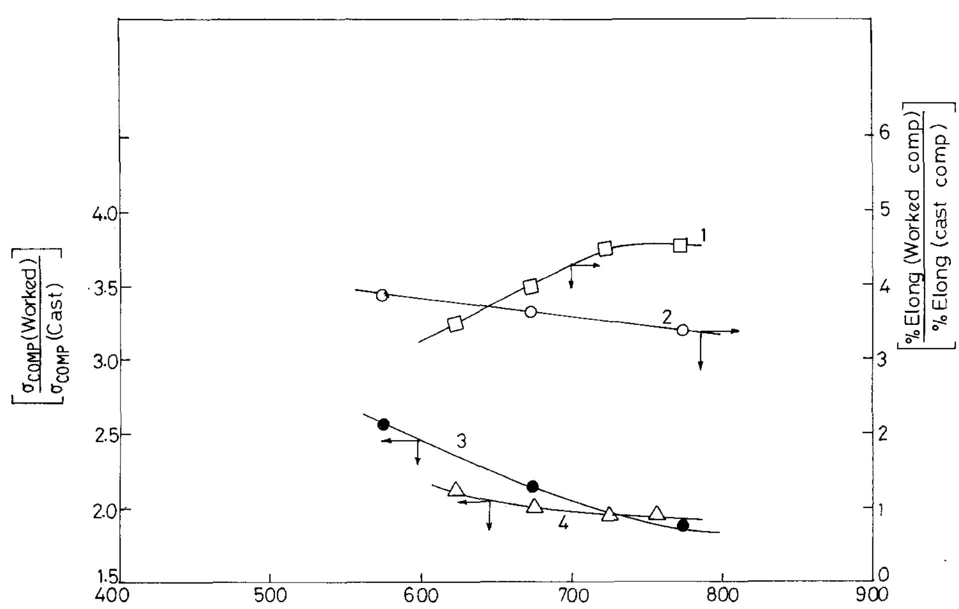


Figure 5 Normalized UTS and per cent elongation of worked composites as a function of extrusion temperature for some aluminium alloy composites (1, 4 [16]; 2, 3 [18]).

composites, Marcus [20] found the UTS of his powder metallurgically-produced SiC-reinforced 6061 aluminium alloy to increase as a result of extrusion. However, this increase was almost half that of SiC whisker-reinforced composite in the as-extruded condition. Auger electron spectroscopy of fractured tensile specimen showed substantially more SiC/Al interfacial failure in the SiC powder–aluminium composite. This difference between particulate- and whisker-reinforced SiC composite could be due to non-uniform sizes and spatial distribution of SiC particles which could have promoted failure in the SiC/Al alloy interface in the case of SiC particle composites. Hot extrusion of magnesium-based alloys, containing SiC whiskers has been reported to result in very large increases in stiffness and strength of these composite materials [21].

Divecha *et al.* [22] have extruded Al–SiC whisker composite fabricated by conventional powder metallurgy techniques. The SiC whiskers typically had a length of 50 μm and a diameter varying between 0.2 and 1 μm. The measured tensile strength of the composite containing 25 vol % whiskers was in the range 552 to 641 MPa whereas the matrix alloy in the T_4 condition had a strength in the range 400 to 435 MPa. In addition to the strengthening due to the presence of SiC whiskers ($\sigma_w = 1.36$ GPa), the strong texture effect and a high dislocation density around the whiskers in the matrix alloy produced additional improvements in the composite tensile strength [23].

3. Effect of processing parameters

Very few systematic studies relating the working process parameters and the corresponding variations of mechanical properties are reported in the literature. In one such study Yuasa and Morooka [18] carried out hot extrusion of Al–Si alloys containing nickel-coated graphite particles in different sizes and volume fractions under various conditions of extrusion temperatures, ram speed, extrusion ratios and die angles. Graphite particles were reported to extend in a fibre-like manner in the plastic deformation zone of the matrix material in a manner similar to that seen in other studies [12] as a result of hot working. The extending ratio (length of graphite after extrusion divided by original diameter before working) tends to decrease with increasing extrusion temperature presumably because the relative plasticity of inclusions decreases with a rise in temperature. The extending ratio also decreases with a decrease in the ram speed during extrusion. Furthermore, this ratio increases and its change with extrusion temperature becomes more pronounced with increase of extrusion ratio. The elongation and UTS increased as a result of hot extrusion. The increase in the tensile strength of the composite due to hot extrusion was found to become more pronounced with increase of ram speed and extrusion temperature. The strength of the alloys containing finer graphite particles becomes comparable with that of graphite-free base alloy extruded at 723 K. The relatively coarser particle dispersions resulted in lower increases in UTS and the strength decreased with a decrease in graphite content [18]. The decrement of UTS per graphite content of unit per cent in the extruded alloys varies from 7.8 MPa with increase of extrusion temperature from 573 to 773 K.

4. Fracture toughness

Fracture toughness of metal–matrix particulate composites is an important property for fail-safe design in structural applications. Relatively little work has been done on the fracture toughness of particulate composites. Divecha *et al.* [22] have reported that the fracture toughness of Al–SiC composites increased from 13 to 34 MPa m$^{1/2}$ as a result of particle addition. Keshavaram *et al.* [24] have reported that hot-pressed Al–2% Mg–8% glass composites had a K_{Ic} value of 16.63 MPa m$^{1/2}$, whereas Al–2 Mg alloy containing 10 wt % flyash had K_{Ic} of 14.76 MPa m$^{1/2}$. In general, glass and flyash particle dispersions were found to decrease the K_{Ic} value of base alloys, possibly because the particles act as potent nucleation sites for voids. Addition of magnesium to these composites was reported [25] to improve the K_{Ic} value. Magnesium inhibits void nucleation by providing improved particle–matrix bonding [26].

Pillai *et al.* [19] studied the fracture toughness of cast and forged Al–3 Mg alloys containing dispersions of graphite and zircon particles. (The introduction of graphite particles decreases the toughness value of the base alloy.) The fracture toughness (CTOD and J) values of the cast and forged aluminium–graphite composites increased with increasing wt % (1.5 to 4.5) of graphite phase. The forged composites were reported to possess higher (70 to 100%) fracture toughness compared to cast composites for a given weight per cent graphite. The CTOD and J toughness values were found to decrease continuously with increasing weight per cent of zircon particles over the range 10 to 30 wt %. This behaviour is similar to that reported by Keshavaram *et al.* [24] for other particles such as glass and flyash. The forged composites generally possessed higher fracture toughness than the cast composite. In both cast and forged aluminium–graphite and aluminium–zircon composites, the fracture toughness decreased with increasing particle size for a given weight per cent of particulates; the decrease in the case of aluminium–zircon was, however, only marginal. The fracture surfaces of forged aluminium–graphite and aluminium–zircon composites showed frequent particle cracking in addition to particle–matrix separation whereas cast composites fractured predominantly by particle–matrix debonding and nucleation of cracks in the matrix. The fracture toughness (CTOD) decreased with increase in interparticle spacing in both the cast and forged aluminium–graphite and aluminium–zircon composites. In the cast zircon composites, CTOD toughness value increases with the interparticle spacing. In the forged condition, the graphite composites showed superior toughness than the zircon composites. K_{Ic} values for Al–3% Mg alloy containing 10 and 15 wt % zircon particles were 24.31 and 24.95 MPa m$^{1/2}$, respectively, in the forged condition. Similarly K_{Ic} values for Al–3% Mg alloy containing 1.5 and 3.0 wt % graphite particles were

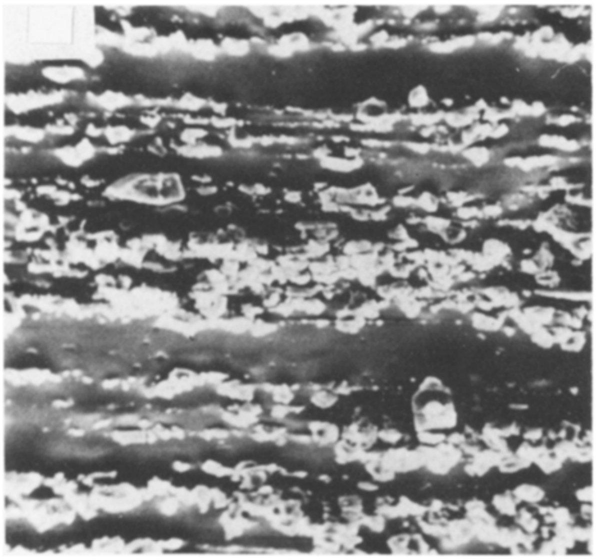


Figure 6 Scanning electron micrograph of extruded Al–4% Cu–0.75% Mg alloy matrix composite containing 15%, 3 μm size alumina particles, × 530 [7].

36.85 and 38.79 MPa m$^{1/2}$, respectively, in the forged condition, and in the cast condition the corresponding values were 26.78 and 28.00 MPa m$^{1/2}$, respectively.

5. Structure and strengthening mechanisms

Fig. 6 shows longitudinal section of extruded Al–4% Cu–0.75% Mg alloy matrix composite containing 15%, 3 μm size alumina particles [7]. The particles dispersed in the matrix tend to align themselves along the direction of extrusion. Fig. 7 shows a discontinuous graphite fibre–aluminium alloy composite where the fibres are also seen to align themselves along the extrusion axis. In this system, extensive formation of brittle aluminium carbide (Al_4C_3) phase at the fibre–matrix interface occurred [7] because of long contact times between partially solid alloy slurry and the fibres during fabrication. In order to obtain effective strengthening, time and temperature during solidification processing must be optimized in a manner so as to give an adequate fibre–matrix bond without excessive reaction at the interface. Because of excessive formation of Al_4C_3 in the Al–alloy–graphite composites no improvement in mechanical properties could be achieved [7] in the above system.

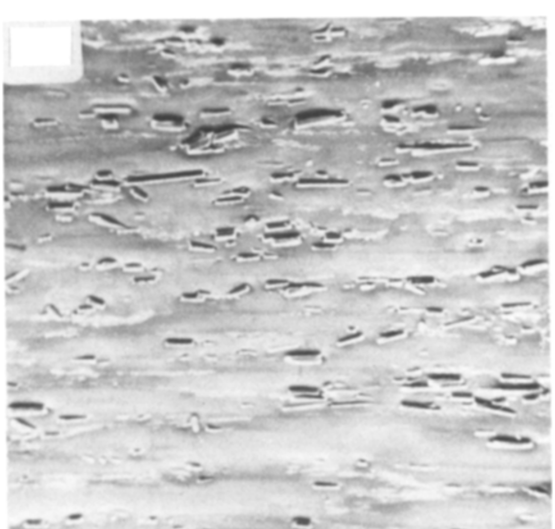


Figure 7 Scanning electron micrograph of extruded 6061 aluminium alloy matrix composite containing 3%, 7 μm diameter discontinuous graphite fibres, × 52 [7].

Several authors [7, 8] have reported the highest increase in yield stress and hardness with smallest particle sizes in hard particle-dispersed alloys. These properties decreased almost linearly with increasing particle sizes [7]. In the presence of a certain volume per cent and size of dispersed non-metallic particles, the hot extruded and annealed composite specimens generally show pronounced textures with accompanying anisotropy of mechanical properties. The strengthening mechanism in these worked materials depends on the volume fraction of particles; at low volume fractions texture hardening makes an overriding contribution to the mechanical properties and at high volume fractions (small interparticle spacing) dispersion hardening chiefly contributes to strengthening [8]. Neuss and Wasserman [27] have reported that extruded composites of aluminium- or copper-based alloys containing plastically non-deformable, non-metallic inorganic particles (e.g. MgO, glass below T_g) reduce or eliminate the known fibre textures during extrusion. On the other hand, layer lattice structures (graphite, MoS_2) influence texture significantly at higher volume fractions of these additives. These workers also noted [27] a large amount (up to 90% reduction in area) of cold drawing of the extruded rods containing non-deformable additives. Additions of a layer lattice structure were reported to reduce the per cent reduction in area more, compared to undeformable materials.

Wasserman [28] has carried out extensive studies on deformation mechanisms and texture strengthening of several alloys containing both deformable and non-deformable additives. All non-deformable particles (e.g. oxides, alumina, silica), particles having only limited plasticity (e.g. nitrides, carbides, salts, silicon and MgO), and easily deformable metallic particles (iron in aluminium or tungsten in copper) having yield strength $\sigma_p \gg \sigma_m$ of the matrix, generally lead to inhibition of texture formation in worked composites.

Soft layer structure materials such as graphite deform by a crystallographic shearing mechanism on the basal plane of the hexagonal crystal. Under combined shear and compressive stresses which exist during thermomechanical processing, plate texture develops because the cohesion between layers normal to the basal plane is low and cleavability is high. The deformation mechanism is here characterized by rotation of the plate-like particles along the direction of deformation. The deformation texture in these cases, therefore, has a basal plane which is parallel to the axis of deformation. The sharpness of the texture of the matrix is most pronounced in the matrix metal without additives, intermediate in alloys containing the layer structure additives and least in alloys containing equivalent amounts of non-deformable particles. The strengthening mechanisms in hot worked composites containing a spectrum of additives with rigid non-deformable particles on one end to high elongated one-dimensional fibres on the other, range from dispersion hardening to fibre reinforcement, strain hardening of the matrix metal to a common strain hardening of both components and finally to *in situ* development of very thin fibres, reaching their near theoretical strength.

Amorphous substances, e.g. glass, behave as non-deformable particles at temperatures below their glass transition temperatures (T_g) whereas they may form fibre texture at temperatures above T_g where they deform by viscous flow. Considerably lower extrusion presssures are required to deform aluminium alloy–glass particle composites because viscous flow of glass during processing provides improved lubrication conditions [9, 11]. At extrusion temperatures the viscosity of the dispersed glass (B_2O_3–based) particles matches the ability of the matrix materials to flow. The strengthening of the worked aluminium–glass composites (Fig. 4) is primarily due to $\langle 1\,1\,1 \rangle$ type texture formation in the matrix around the glass fibres as well as the formation of high dislocation density zone and/or finer subgrain structure around the glass strands [9].

In the hot worked samples of Al–1% Mg–graphite particle composites, graphite particles are deformed in a stringer-like manner oriented parallel to the working direction [16] as shown in Fig. 8. The stringers are non-uniform in cross section and have a range of length to diameter ratios with a maximum value of $l/d = 20$ [14], whereas before extrusion this value is usually 1 or 2. Phenomenologically, the deformation of graphite into stringers as a result of working can be explained by invoking the model of Gore and Charles [29] according to which second-phase particles are deformed only if the ratio of the hardness of the particles to that of the matrix is less than a critical value, i.e.

$$v = a - \frac{H_i}{H_m} \qquad (1)$$

where v is the relative plasticity, H_i, the initial hardness of second phase particles, H_m the initial hardness of the matrix and a is a constant equal to 2.0 when inclusions are not well bonded to the matrix and it may be as high as 6 for well-bonded inclusion. During extrusion of the composites, the hardness ratio of graphite to aluminium is 0.4 at room temperature.

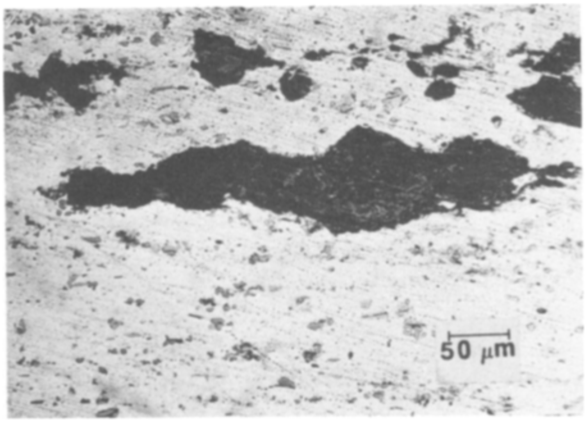


Figure 8 Micrograph showing a graphite stringer aligned parallel to extrusion direction in hot-extruded ($T = 673$ K, extrusion ratio $R = 9.77$) Al–1% Mg–2.5% graphite particle composite.

Even at the processing temperature this ratio is expected to attain a maximum value of 1.5 so that v is still positive. Graphite is generally brittle under tensile loading both at room temperature and at 673 K because graphite cannot be deformed by a simple shear on the basal plane (even though it is the slip plane) unless a compressive stress acts normal to the slip plane. During extrusion a combination of shear and compressive stresses exists and makes the deformation of graphite into stringers possible. However, because the distribution of the shear and compressive components is expected to be non-uniform from one particle to another depending upon their location, morphology and orientation, the graphite stringers cleavage fracture as shown in Fig. 12a in graphite-dispersed aluminium-alloys after tensile testing.

A fine cup and cone structure was also observed [18] at the fracture surface indicating that the composites fracture in a ductile manner. The graphite containing alloys were, however, found to fracture in a brittle manner in the as-cast condition (Fig. 12b).

The improvement in strength and ductility as a result of thermomechanical processing of aluminium alloys containing layer structure particles appears to be due to:

1. work hardening of the matrix;
2. refinement of cast structure (e.g. the dendritic structure, Fig. 13, of Al–1% Mg–2.5% graphite alloy was completely eliminated [15] after working, Fig. 8;
3. extension of graphite particles in a stringer-like manner along the working direction;
4. improved metal matrix–particle bonding;
5. decreased porosity level in the worked composite (indicated by an increase in the density after working), and
6. texture hardening.

In view of the increasing potential of technological pay-offs (high wear resistance, high strength) of hot working of metal–matrix particulate composites, there is a need for systematic investigation of the strengthening mechanism in these materials and their relation to processing parameters, structures, and particle size, morphology and distribution. This review has served to summarize the limited knowledge currently available in the area of thermomechanical processing of cast metal–matrix particulate composites with a view to generating further interest in this potentially significant area.

6. Conclusions

This paper analysing the thermomechanical processing of cast aluminium-based particulate composite indicates that there is considerable potential in enhancing the properties of cast composites through thermomechanical processing. The cast versions of these composites exhibited excellent tribological properties, even though they exhibited marginal but adequate mechanical properties. The ability to enhance their mechanical properties through thermomechanical processing without impairing their tribological properties, would increase their potential use in severe tribological applications. In addition, the increases in strengths, modulus and fracture toughness through

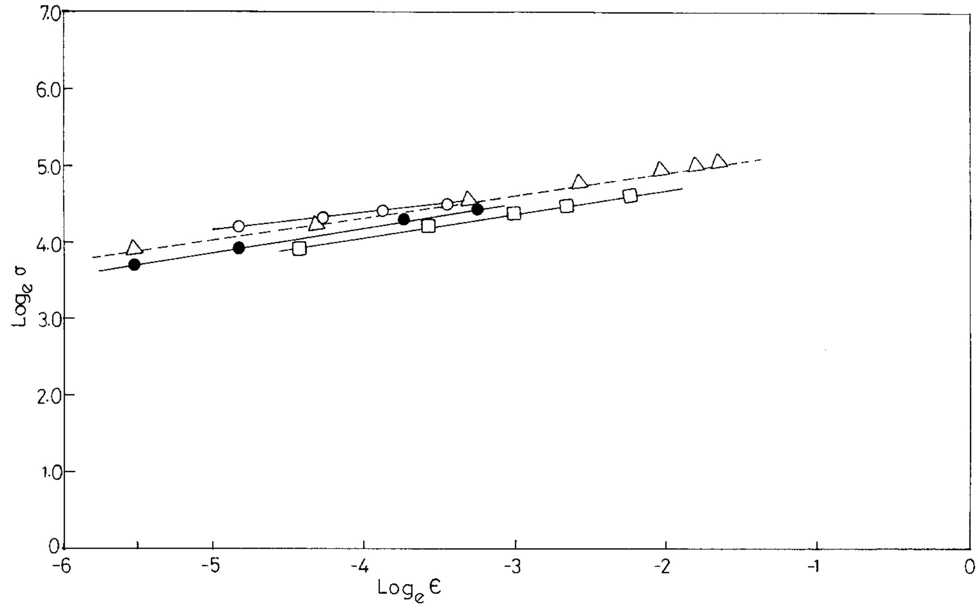


Figure 9 Ln σ plotted against ln ε for Al–1% Mg, Al–1% Mg–2.5% graphite (cast and extruded) and Al–3% Al_2O_3 (cast) particulate composites. Log σ = n log ε + log k. (○) Al–3% Al_2O_3 cast [31], n = 0.242, K = 200 MPa; (●) Al–1% Mg–2.5% graphite (cast) [16]. n = 0.329, K = 233 MPa; (△) Al–1% Mg–2.5% graphite (extruded) [16], n = 0.293 K, K = 245 MPa; (□) Al–1% Mg (cast) [16], n = 0.317, K = 288 MPa.

proper thermomechanical processing may make them competitive in high-strength high-stiffness structural applications.

Much of the work of thermomechanical processing of cast composites involves random probing to demonstrate the potential of increasing mechanical properties. Considerable systematic analytical and experimental work is necessary to develop predictive capability in this area, and to explore fully the potential of these attain non-uniform cross-sections in the extruded samples.

The strain-hardening behaviour of both cast and worked composites was analysed by using the Hollomon equation [30] $\sigma = K\varepsilon^n$ where K is the strength coefficient and n is the strain-hardening exponent. In the case of cast Al–Al_2O_3 composites the equation

$$\ln \sigma = \ln K + n \ln \varepsilon \qquad (2)$$

describes the work-hardening behaviour with n = 0.242 and K = 200 MPa. The data for hot-worked Al–Al_2O_3 composite is, however, not available and a comparison of strain-hardening behaviour in cast and worked composites was not possible in this system. For Al–1% Mg alloy containing 2.5 wt % graphite particles (Fig. 9) a single value of exponent n describes the strain-hardening behaviour in both as-cast and extruded conditions. The value of n decreased from 0.329 to 0.293 as a result of hot working, whereas the strength coefficient increased from 233 to 245 MPa. The effects of dispersoid weight fraction on the strain-hardening behaviour of cast and worked Al–3% Mg alloy containing different quantities of graphite and zircon particles are shown in Figs 10 and 11, respectively. In general, departure from linearity characterized ln σ against ln ε plots for these composites in cast and worked conditions so that more than one value of strain-hardening exponent seems to e operative. At

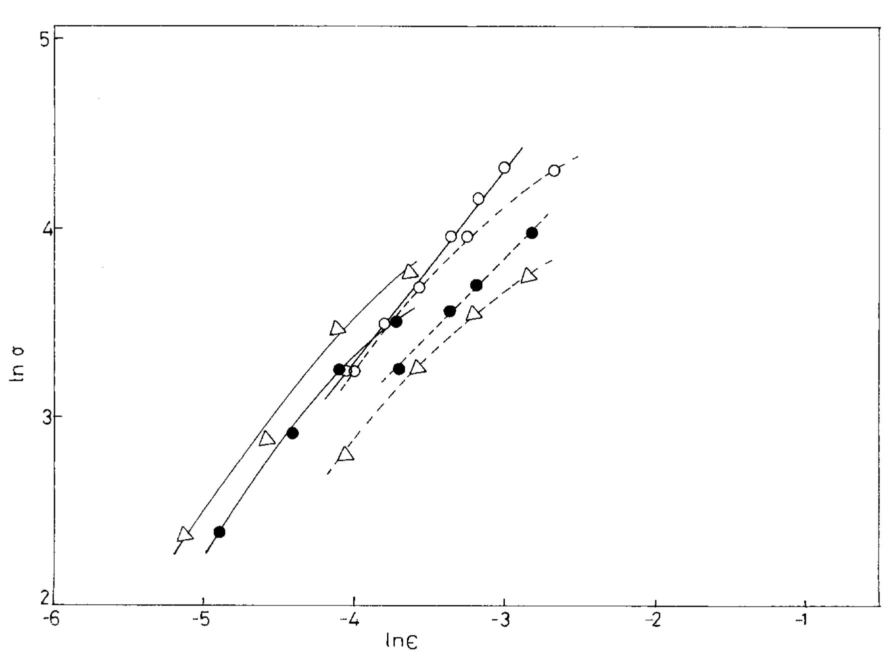


Figure 10 Ln σ plotted against ln ε for Al–3% Mg–graphite particle composites for different graphite contents in (– – –) cast and (——) forged conditions. (○) Al–3% Mg–1.5% graphite [19], (●) Al–3% Mg–3% graphite [19], (△) Al–3% Mg–4.5% graphite [19].

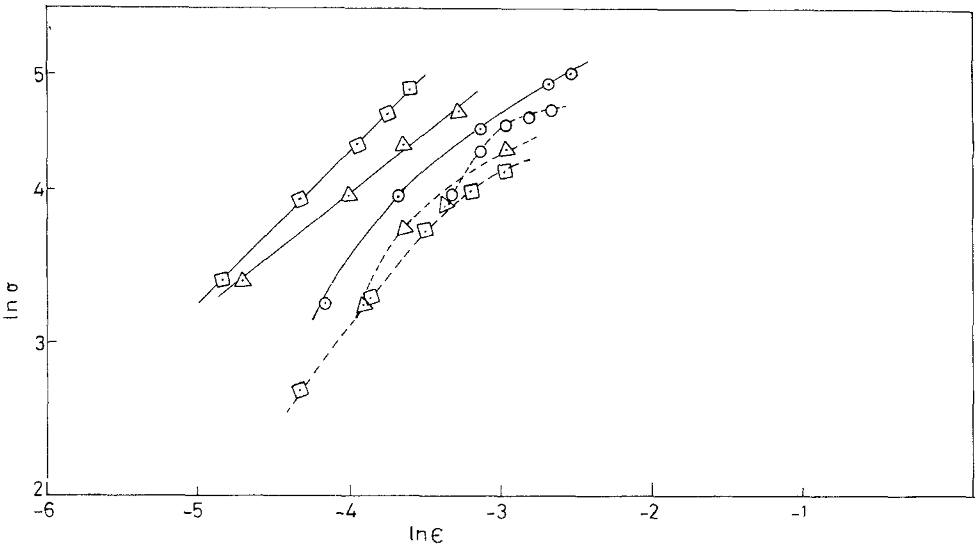


Figure 11 Ln σ plotted against ln ε for Al–3% Mg–zircon particle composites for different zircon contents in (– – –) cast and (——) forged conditions. (⊙) Al–3% Mg–10% zircon [19], (△) Al–3% Mg–20% zircon [19], (⊡) Al–3% Mg–30% zircon [19].

higher weight per cent zircon in Al–3% Mg alloys, however, the plots of ln σ against ln ε showed linear behaviour in the forged condition. The mechanisms of strain-hardening behaviour operative under these conditions are yet to be elucidated.

The density of the cast Al–1% Mg–2.5% graphite composite increased from 2.64 g cm^{-3} in the as-cast condition to 2.68 g cm^{-3} in the extruded condition, indicating elimination of porosity after working.

Generally no cracking in the matrix material around graphite particles was observed [18] as a result of working. The extended graphite particles serve to reduce the notch effect against fracture in the matrix material. Extruded rods of aluminium alloy-graphite particle composite show die lines on the surface on visual examination [15, 18]. These defects were more numerous in rods that were extruded at higher temperatures [15] of 723 and 773 K using an extrusion ratio of 25. Yuasa and Morooka [18] reported the presence of die lines on both graphite-free and graphite-containing Al–Si alloys, indicating that these defects are characteristic of the material itself and are not caused by the presence of graphite particles.

Fractography [16] of extruded, and rolled and forged composites of aluminium–graphite showed materials. The studies to data indicate that, in general, composites with finer dispersoids which deform and fibrize show maximum enhancements in properties as a result of thermomechanical processing.

Dispersoids which have layered structures form stringers and lead to texturing and greater enhancement of properties. Thermomechanical processing which seals porosity in cast composites leads to substantial enhancements in properties, especially ductility. The relative plasticity of the dispersoids and the matrix governs whether the dispersoids deform, and therefore the selection of deformation temperatures is quite critical in its influence on enhancements of properties. Combinations of properties of matrix and dispersoids, and the thermomechanical processing sequence, should be such as to leave high dislocation densities in the matrix, near the interface, because this leads to greater enhancements in strengths. The bonding between the matrix and the interface should be strong, before and during thermomechanical processing, to maximize the increases in strength; segregation of certain elements at the interface which enhance bonding seem to improve the strength of composites. If the volume and shape of the dispersoids, the mean free paths and the differences in coefficients of thermal expansion are selected optimally, the matrix undergoes constrained deformation, has high dislocation density with possible residual compressive stresses resulting in improved properties. In certain instances thermomechanical

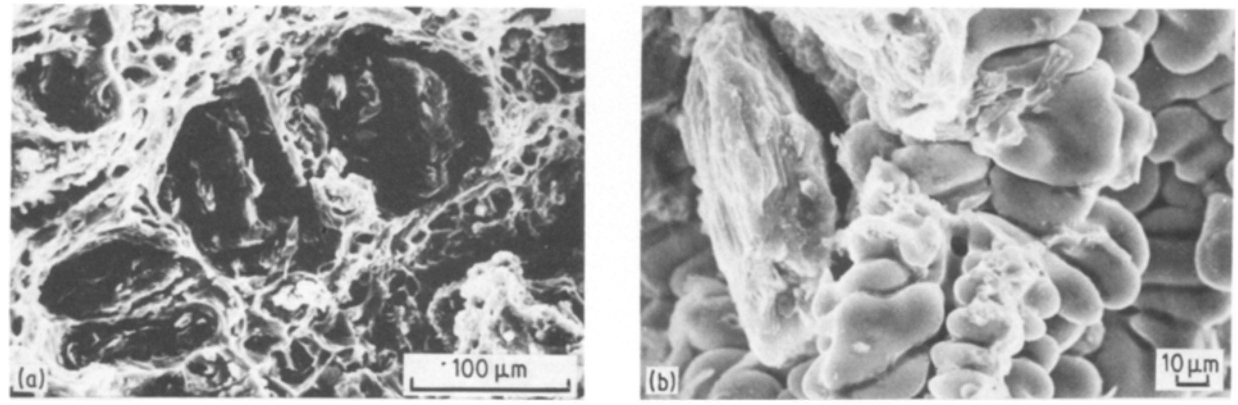


Figure 12 Scanning electron fractographs showing (a) cleavage fracture in hot-extruded aluminium alloy–graphite particle composite, (b) brittle fracture in the as-cast composite of (a).

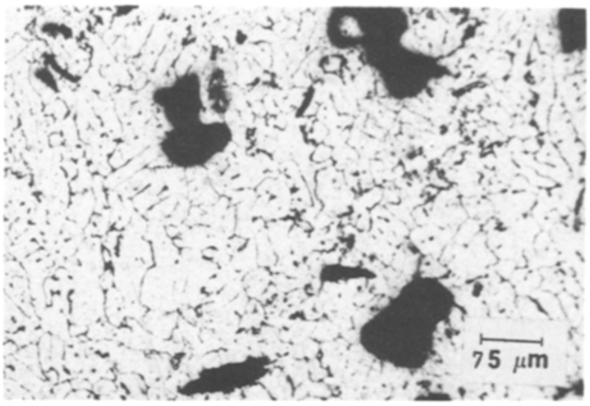


Figure 13 Micrograph showing dendritic structure in Al–1% Mg–2.5% graphite composite prior to extrusion.

processing can be selectively used to fragment the dispersoids into finer sizes and redistribute them with fresh surfaces in contact with the matrix leading to improvements in properties. These are a few generalized observations concerning thermomechanical processing of metal–matrix composites; the specific thermomechanical cycle selected will depend upon the particular composite in question, and the property targets.

Acknowledgements

The authors gratefully acknowledge the assistance of Mr K. K. S. Gautam in SEM work, Mr N. Saha for line drawings and Mr Madhav Lal for meticulous typing of the manuscript.

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Received 6 October 1986
and accepted 15 May 1987