JOURNAL OF MATERIALS SCIENCE 30 (1995) 2269–2276

Ageing and toughness of silicon carbide particulate reinforced Al–Cu and Al–Cu–Mg based metal–matrix composites

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Metal–matrix composites (MMC) comprising powder aluminium alloys reinforced by particulate ceramic are being developed for widespread aerospace structural applications ranging from fuselage and missile components to undercarriage parts. Most interest is centred on MMCs with an Al–Cu–Mg (2 1 2 4) matrix alloy. These MMCs possess high levels of specific stiffness with high specific strengths but can exhibit lower ductility and toughness than conventional unreinforced aluminium alloys.

To overcome these problems the effects of alloy composition on the ageing behaviour and notched tensile properties of Al–Cu–Mg and Al–Cu based alloys reinforced with 20 wt% silicon carbide particulate have been investigated.

Al–Cu–Mg MMCs gave higher strengths and moduli than unreinforced sheet. Lowering the copper and magnesium content resulted in reduced strength but did not affect the rate of age hardening. The Al–Cu MMCs gave the lowest strengths but the absence of natural ageing may prove advantageous, enabling sheet to be formed and subsequently heat-treated to the peak strength condition.

1. Introduction

Metal–matrix composites (MMCs) which comprise an aluminium alloy reinforced with a ceramic material in particulate form and produced by a powder metallurgy route are currently being developed for structural aerospace applications where high specific strengths and stiffnesses are required. Potential components that could utilize MMCs include missile bodies and fins, aircraft floor support struts and beams, and undercarriage components.

Al–Cu–Mg alloys are currently being investigated as a matrix for silicon carbide particulate reinforced metal–matrix composites (MMCs). This type of MMC typically contains 20 wt % of the particulate reinforcement which results in significant increases in specific stiffness and specific strength in comparison with conventional aluminium alloys. However the reinforcement additions have an adverse effect on toughness and ductility [1]. The composites are also thought to exhibit a rapid natural ageing response after solution heat treatment which can lead to difficulties when product forms such as extrusions and sheet forming or stretching after solution heat treatment. Consequently optimization of the composition of the aluminium matrix alloy is required if this type of MMC is to be used in significant amounts for aerospace components. With this objective in mind, a powder metallurgy production route has been set up at DRA Farnborough involving the use of an inert gas molten metal atomizer. This has enabled various MMCs containing Al–Cu and Al–Cu–Mg based alloys to be made and the fabricability of the materials to sheet form to be assessed [2]. The effects of alloy composition and ageing conditions on the microstructures and mechanical properties of the MMCs have also been studied with notched tensile tests used to assess the toughness of the materials.

2. Material and experimental procedures

MMC billets 125 mm long × 55 mm diameter were manufactured using a powder metallurgy route described in detail elsewhere [2]. The billets contained 20 wt% silicon carbide particulate (average particle size of 3 μm)* and aluminium alloy powder either made using the inert gas atomization facility at DRA Farnborough or commercially produced by Metalloys Ltd. Unreinforced billets were also made for comparison purposes. The DRA powders were obtained from pre-alloyed ingots which were melted and superheated to 850 °C before being atomized using either argon or an argon–3% oxygen gas mixture. The compositions of the alloys are given in Table I.

After atomization, the powders were blended with dry, sieved silicon carbide and cold compacted into aluminium cans which were fitted with lids (with 2 mm diameter degassing holes) by electron beam welding. Degassing was done within a furnace inside the

*Supplied as F1200 grade grit by Sohio Electro Minerals Ltd.

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TABLE I Composition of alloy matrix

Alloy	Composition in wt %	
1	2Cu–1Mg–0.6Mn	Remainder Al
2	2Cu–1Mg–0.12Zr	Remainder Al
3	4.35Cu–1.5Mg–0.6Mn	Remainder Al
4	4Cu–1.7Mg–0.8Mn[a]	Remainder Al
5	4.45Cu–1.5Mg–0.13Zr	Remainder Al
6	4.35Cu	Remainder Al
7	4.7Cu–0.12Zr	Remainder Al
8	6Cu	Remainder Ai

[a] 2124 alloy powder commercially atomized

electron beam welding chamber which was closed and evacuated to a pressure of typically 2×10^{-6} mbar. The cans were heated under vacuum to temperatures between 500 and 530 °C, held at temperature for 1 h and allowed to cool to ~ 300 °C prior to sealing the degassing hole by electron beam welding. The billets were consolidated by hot isostatic pressing (HIP'ing) for 1 h at a pressure of 2500 bar using the same temperatures as were used for degassing. The aluminium cans were then removed by machining and the billets pre-heated for 1 h at 475 °C, prior to forging diametrically to 33 mm, reheating for 30 min and forging to 22 mm. The plates were finally hot rolled at 475 °C to a sheet thickness of 2 mm using reductions of 10–15% per pass.

Solution heat treatments (SHT) were carried out for 30 min at either 505 °C or 530 °C in an air circulating furnace followed by quenching in cold water. Samples were artificially aged at 150 °C for times up to 1650 h.

Plain tensile testpieces, and notched tensile testpieces [3] were machined parallel to the sheet rolling direction. The notch-yield ratio (i.e. the ratio of the notched tensile strength to the 0.2% proof stress) was used as a measure of the toughness of samples. Tensile testing was carried out to EN2002/1 (formerly BS18 cat2) using a transducer extensometer of gauge length 20 mm.

Metallographic specimens were prepared by grinding with carborundum papers, polishing to 0.25 μm diamond using napless cloths and finally polishing with a proprietary alumina based suspension.

Thin foils for transmission electron microscopy were prepared from 5 mm discs punched out from sheet specimens which had been ground to within 0.2 mm of the mid-thickness using carborundum papers. The discs were electropolished in a solution containing 30% nitric acid and 70% methanol at − 30 °C using a twin-jet polishing machine. The foils were examined in a Jeol 200CX transmission electron microscope.

3. Results

3.1. Effects of natural ageing on tensile properties

The effects of natural ageing on the tensile properties of the unreinforced Al–Cu–Mg alloy (alloy 4) and the reinforced Al–Cu–Mg matrix alloys (alloys 1 to 5) after SHT at 505 °C are shown in Figs 1 and 2. The

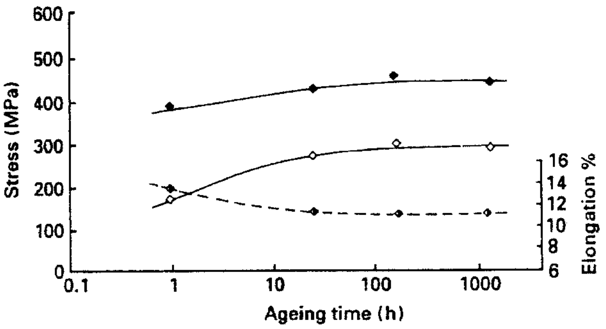

Figure 1 Effect of solution heat treatment at 505 °C, cold water quenching followed by natural ageing on the tensile properties of unreinforced Al–4%Cu–1.7%Mg–0.8%Mn sheet. ◆ Tensile strength; ◇ 0.2% PS; ⬖ ductility.

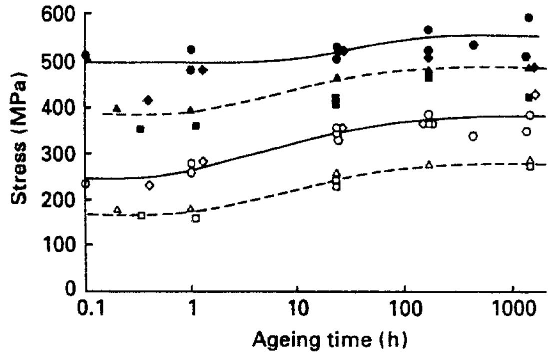

Figure 2 Effect of alloy composition and natural ageing on the tensile properties of reinforced Al–Cu–Mg alloy sheet after solution heat treatment at 505 °C followed by cold water quenching. Closed symbols, tensile strength; open symbols, 0.2% proof stress. Solid line, 4% Cu; dashed line 2% Cu. Alloy composition; □, ■ Al–2 Cu–1Mg–0.12Zr; △, ▲ Al–2Cu–1Mg– 0.6Mn; ◇, ◆ Al–4Cu–1.7 Mg–0.8Mn; ○, ● Al–4.35Cu–1.5Mg–0.6Mn; ⬡, ⬢ Al–4.45Cu–1.5 Mg–0.13Zr.

peak strength of the unreinforced alloy was reached after ~ 144 h, the values of 0.2% proof stress and tensile strength being 295 and 450 MPa respectively. During natural ageing the ductility decreased slightly from ~ 13.6 to 11.2%.

The tensile strengths of the reinforced Al–Cu–Mg alloys 3–5, containing nominally 4–4.5% copper and 1.5–1.7% magnesium (Fig. 2), were similar and increased at a similar rate as for the unreinforced alloy. The 0.2% proof stress and the tensile strength after ageing for 1 h were both 100 MPa higher than for the unreinforced alloy (alloy 4), and, on further ageing, increased by another 100 MPa reaching peak values after ~ 144 h. Ductilities were variable particularly in the peak aged condition ranging from 1.9 to 10.6%. The elastic moduli appeared to increase from ~ 83 GPa immediately after quenching to 90–110 GPa in the peak strength condition. Lowering the copper and magnesium contents to 2 and 1% respectively (reinforced alloys 1 and 2) resulted in similar ageing behaviour although the 0.2% proof stresses and tensile strengths in the peak aged condition were lower by 100 and 70 MPa respectively. Ductilities tended to be high (11–14%) during ageing up to 24 h but then decreased to 5.5–10% after 1440 h. The presence of manganese or zirconium in the reinforced

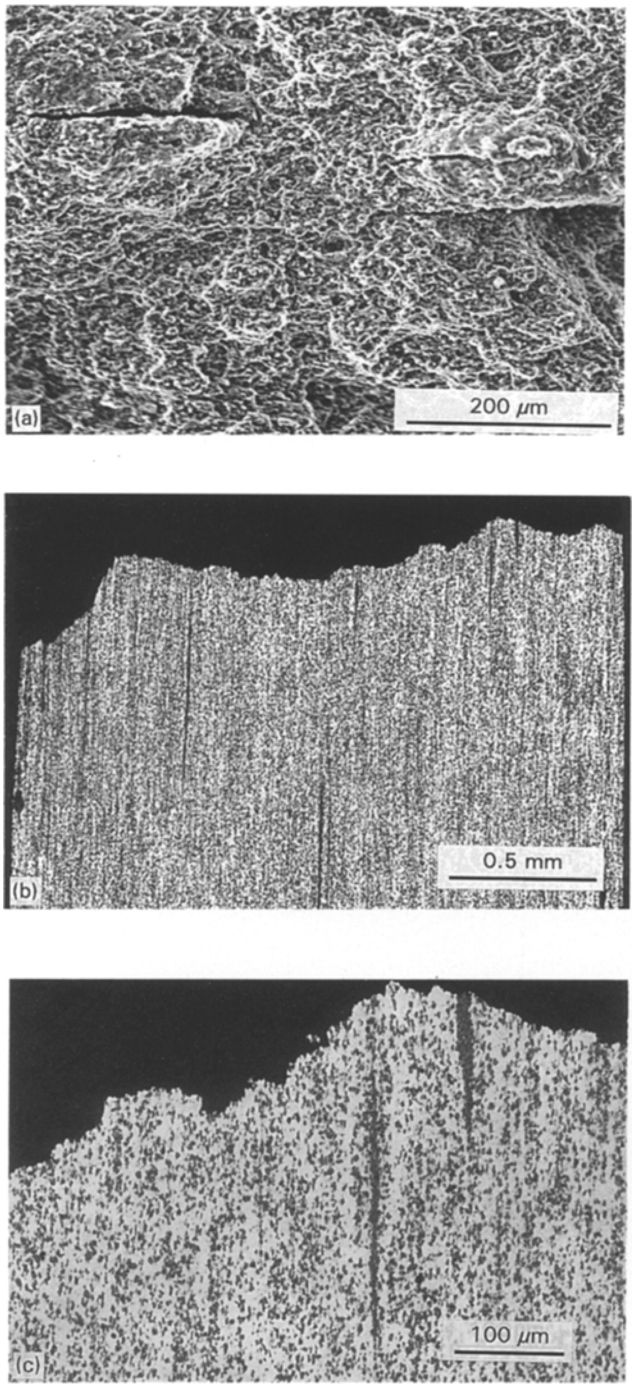

Figure 3 Fracture of reinforced Al–4%Cu–1.7%Mg–0.8%Mn sheet tested after solution heat treatment at 505°C, cold water quenching and ageing for 24 h at 150 °C (2.2% elongation. (a) localized delamination on fracture surface, (b) and (c) microsections through fracture showing elongated SiC agglomerates.

alloys seemed to have no effect on the tensile properties. It was noticeable that the reinforced Al–4Cu–1.7Mg–0.8Mn (alloy 4) manufactured from commercially produced powder gave low ductilities (2–4.5%) which was attributed to the presence of localized concentrations or agglomerates of silicon carbide particles which resulted in localized delamination of testpieces and consequently premature failure (Fig. 3).

The effects of natural ageing on the tensile properties of the reinforced Al–4.35%Cu and Al–4.7%Cu–0.12%Zr alloys after SHT at 505 and 530 °C respec-

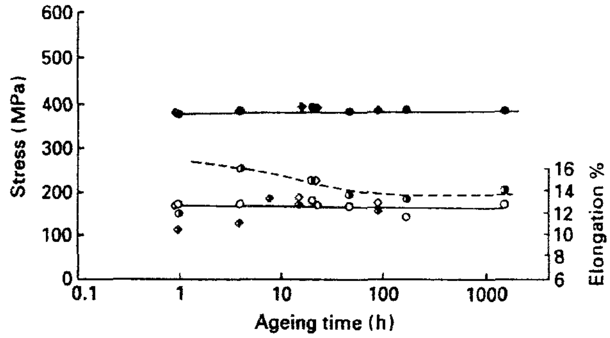

Figure 4 Effect of natural ageing on the tensile properties of reinforced Al–4.35%Cu alloy (○ 0.2%PS, ● TS, ◑ EL%) and Al–4.7%Cu–0.12%Zr alloy (◇ 0.2%PS, ◆ TS, ◈ EL%) sheet, solution heat treatment at 505 and 530 °C respectively followed by cold water quenching. Top curve, tensile strength; middle curve, ductility; lower curve, 0.2% proof stress.

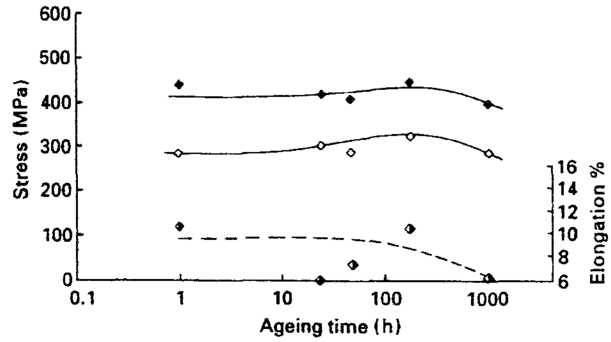

Figure 5 Effect of ageing at 150 °C on the tensile properties of unreinforced Al–4%Cu–1.7%Mg–0.8%Mn alloy sheet after solution heat treatment at 505°C followed by cold water quenching. ◆ Tensile strength; ◇ 0.2%PS; ◈ ductility.

tively are shown in Fig. 4. There were no changes in 0.2% proof stress or tensile strength during natural ageing for times up to 1500 h. The ductilities were significantly higher than for the Al–Cu–Mg alloys with the same copper level (14–16% compared with 8–12%) and only decreased slightly during natural ageing. The zirconium addition to alloy 7 appeared to have no effect on the 0.2% proof stress or tensile strength but the ductilities tended to be reduced slightly.

3.2. Effects of artificial ageing on tensile properties

The effects of ageing at 150 °C on the tensile properties of the unreinforced Al–4%Cu–1.7%Mg–0.8Mn alloy (alloy 4) and the reinforced Al–Cu–Mg alloys 1–5 are shown in Figs 5 and 6. For the unreinforced alloy the 0.2% proof stresses and tensile strengths increased slightly after 48 h, the maximum values after ageing being 320 and 450 MPa respectively. There was some scatter in the ductilities which ranged from 4.5 to 11%, some of the low values coming from testpieces that had delaminated. This mode of fracture was probably associated with entrapped oxides [2] which had formed during processing and impeded the bonding of the powder particles. The 0.2% proof stresses for the reinforced Al–Cu–Mg samples were more sensitive to ageing and increased after 24 h to reach a peak after 120 h. The apparent insensitivity of the tensile

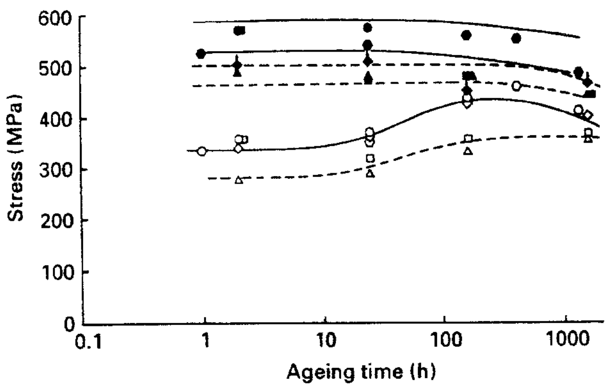

Figure 6 Effect of alloy composition and ageing at 150 °C on the tensile properties of reinforced Al–Cu–Mg alloy sheet after solution heat treatment at 505 °C followed by cold water quenching. Solid lines, 4% Cu; dashed lines, 2% Cu. Open symbols, 0.2% PS; closed symbols, tensile strength. See Fig. 2 for key to symbols.

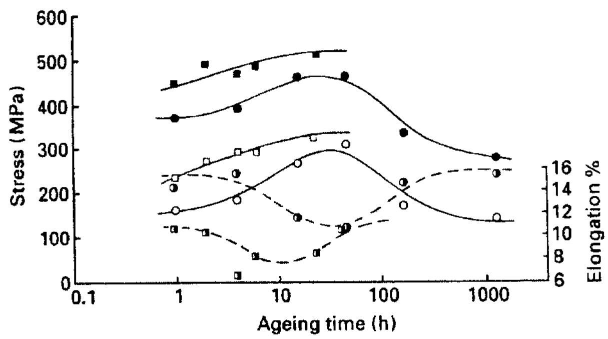

Figure 8 Effect of copper content and ageing at 150 °C on the tensile properties of reinforced Al–Cu alloy sheet. Al–4.35Cu; ○ 0.2% PS, ● TS, ◑ EL%, Al–6Cu; □ 0.2% PS, ■ TS, ◨ EL%.

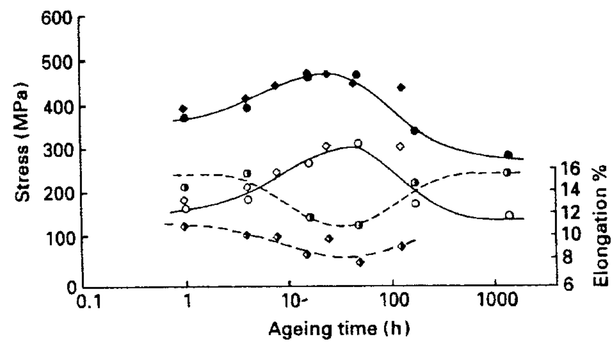

Figure 7 Effect of ageing at 150 °C on the tensile proerties of reinforced Al–4.35%Cu (○ 0.2%PS, ● TS, ◑ EL%) and Al–4.7%Cu–0.12%Zr (◇ 0.2%PS, ◆ TS, ◈ EL%) alloy sheet after solution heat treatment at 505 and 530 °C respectively followed by cold water quenching.

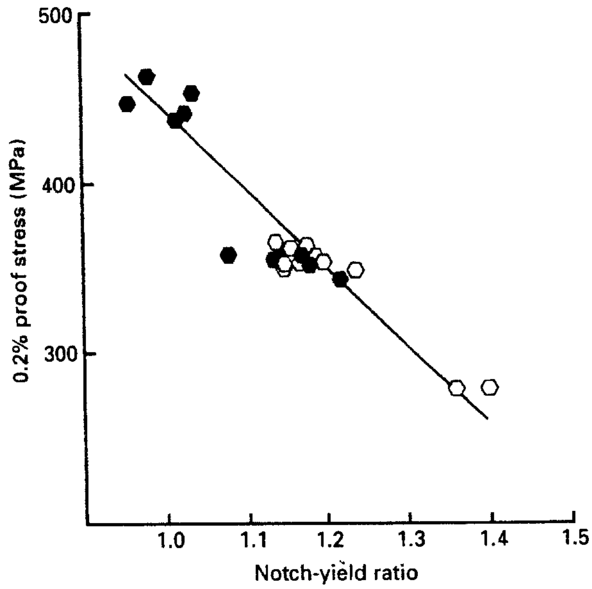

Figure 9 Effect of natural and artificial ageing time on the notch-yield ratio of reinforced Al–4.45%Cu–1.5%Mg–0.13%Zr alloy sheet after solution heat treatment at 505 °C and cold water quenching. ○ Naturally aged; ● artificially aged at 150 °C.

strengths to ageing was related to the presence of materials defects as discussed earlier and was the main reason for the low tensile strengths obtained for the reinforced alloy 4 samples. Reducing the copper and magnesium contents to 2 and 1% respectively decreased the peak aged tensile strengths by 60–80 MPa although after ageing for ~ 1600 h this difference was reduced. With the exception of the reinforced alloy 4 samples which gave low values of ductility (4%), the Al–Cu–Mg samples generally showed ductilities of 8–11% decreasing to 4–7% after ageing for 1600 h.

The effects of ageing on the tensile properties of the reinforced Al–Cu samples (alloys 6–8) are shown in Figs 7 and 8. After short ageing times the 0.2% proof stresses and tensile strengths of the Al–4.35%Cu sample were relatively low at 160 and 370 MPa respectively but reached peak values of 305 and 470 MPa respectively after 24–48 h ageing. Ductilities were relatively high although they decreased to a minimum of 10.7% in the peak aged condition. The sample containing zirconium showed a similar peak strength which was retained for longer ageing times (up to 124 h) but the ductilities were significantly reduced reaching a minimum of 7.6% in the peak aged condition. Increasing the copper content to ~ 6% resulted in higher strength when the sheet was SHT at 530 °C prior to ageing with peak 0.2% proof stress and tensile strength values of 328 and 511 MPa respectively.

3.3. Effects of ageing on the notched tensile behaviour

Owing to natural ageing the notch-yield ratio of reinforced Al–4.45Cu–1.5Mg–0.13Zr alloy decreased from about 1.4 after 1 h to reach a plateau with a ratio of 1.1–1.2 after 24 h. In contrast, ageing at 150 °C resulted in a lower ratio which decreased from 1.0–1.2 after 1 h to values ranging from 0.8–1.0 after ageing for > 168 h. An inverse linear relationship was obtained between the 0.2% proof stresses and the notch-yield ratios with the notch-yield ratio decreasing as the 0.2% proof stress increased (Fig. 9).

Comparison of the notch-yield behaviour of MMCs with different matrix alloy compositions after artificial ageing for different times revealed that the majority of the data points appear to fall on a common line irrespective of the alloy compostion (Fig. 10). Magne-

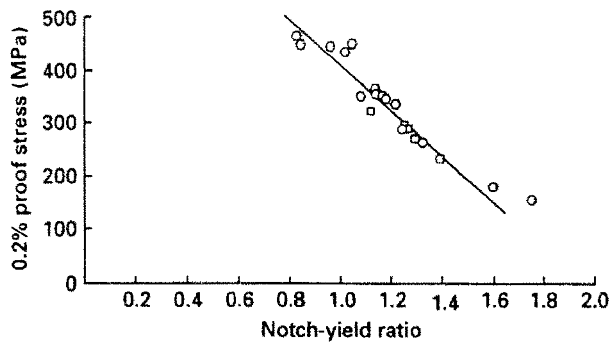

Figure 10 Effect of alloy composition on the notched tensile behaviour of reinforced Al–Cu–Mg and Al–Cu alloy sheets after artificial ageing at 150 °C. ○ Al–4.35Cu; □ Al–6Cu; ⬡Al–4.45 Cu–1.5Mg–0.13 Zr.

sium-free samples showed higher notch-yield ratios indicating higher toughness although the strengths were lower than Al–Cu–Mg MMCs.

3.4. Effects of solution heat treatment on grain structures

The effects of SHT on the microstructures of hot-rolled unreinforced and reinforced sheet are shown in Fig. 11(a–d). The grain structures produced in the unreinforced manganese and zirconium containing Al–Cu–Mg alloys (alloys 4 and 5) were similar and exhibited coarse recrystallized grains with sub-grains typically 50–200 μm in size evident in the L–T plane. The grain structures of reinforced samples were more difficult to distinguish, however, comparison of the microstructures of the reinforced Al–Cu–Mg alloy variants (alloys 3 and 5) revealed that the manganese addition (0.6%Mn) had resulted in a finer grain structure (~ 50 μm in size) than the zirconium addition (0.12%Zr) (compare Fig. 11(c) and (d)).

The grain structures of the reinforced Al–Cu alloys (alloys 6–8) were similar to each other but were coarser than the reinfored Al–Cu–Mg alloy variants,

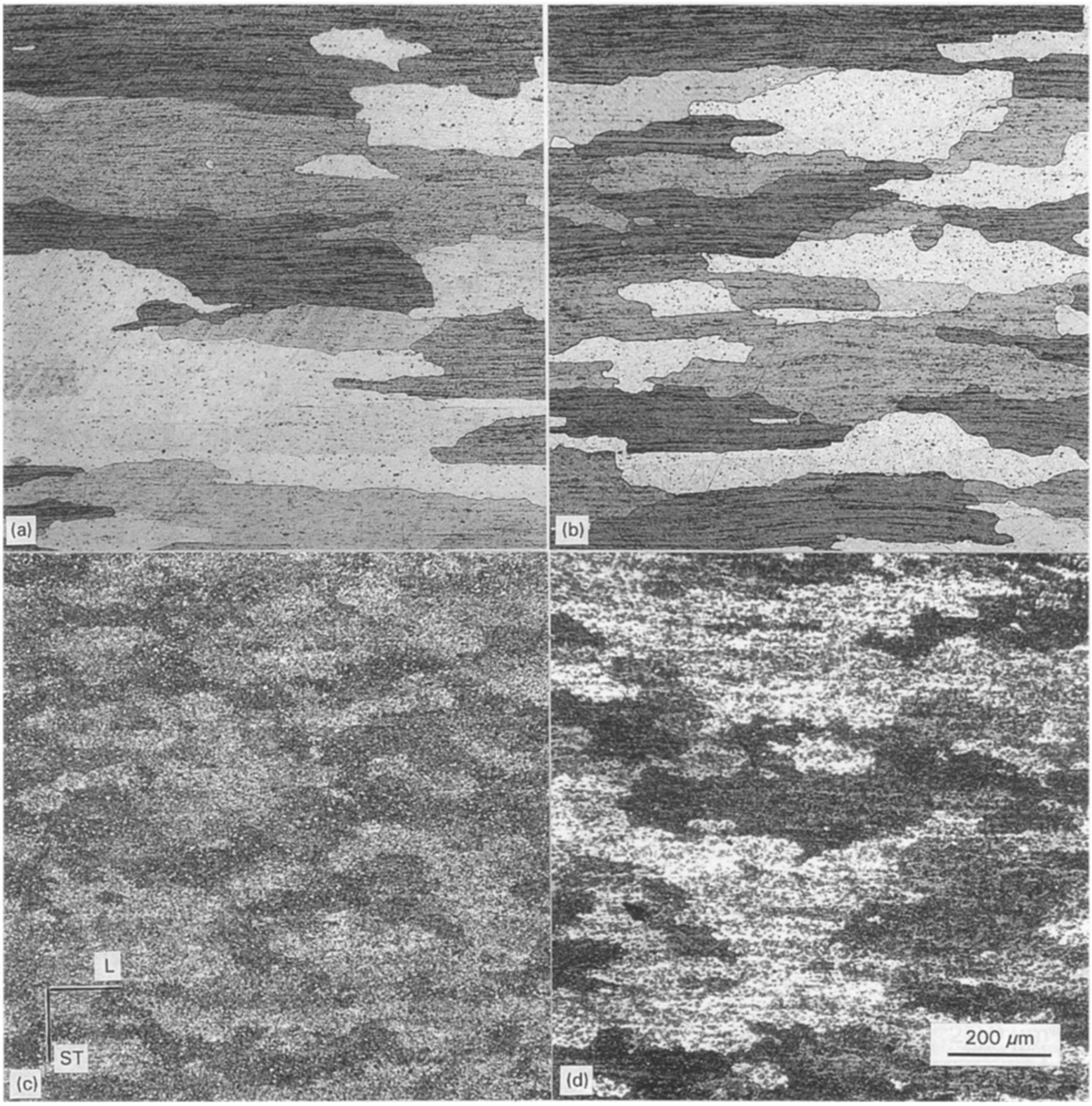

Figure 11 Effect of solution heat treatment for 30 min at 505 °C and cold water quenching on the microstructure of hot rolled sheet; (a) unreinforced Al–4%Cu–1.7%Mg–0.8%Mn alloy; (b) unreinforced Al–4.45%Cu–1.5%Mg–0.13%Zr alloy; (c) reinforced Al–4%Cu–1.7%Mg–0.8%Mn alloy; and (d) reinforced Al–4.45%Cu–1.5%Mg–0.13%Zr alloy.

Figure 12 Effect of solution heat treatment on the microstructure of reinforced hot rolled sheet; (a) Al–4.35%Cu; (b) Al–4.7%Cu–0.12%Zr; and (c) Al–6%Cu.

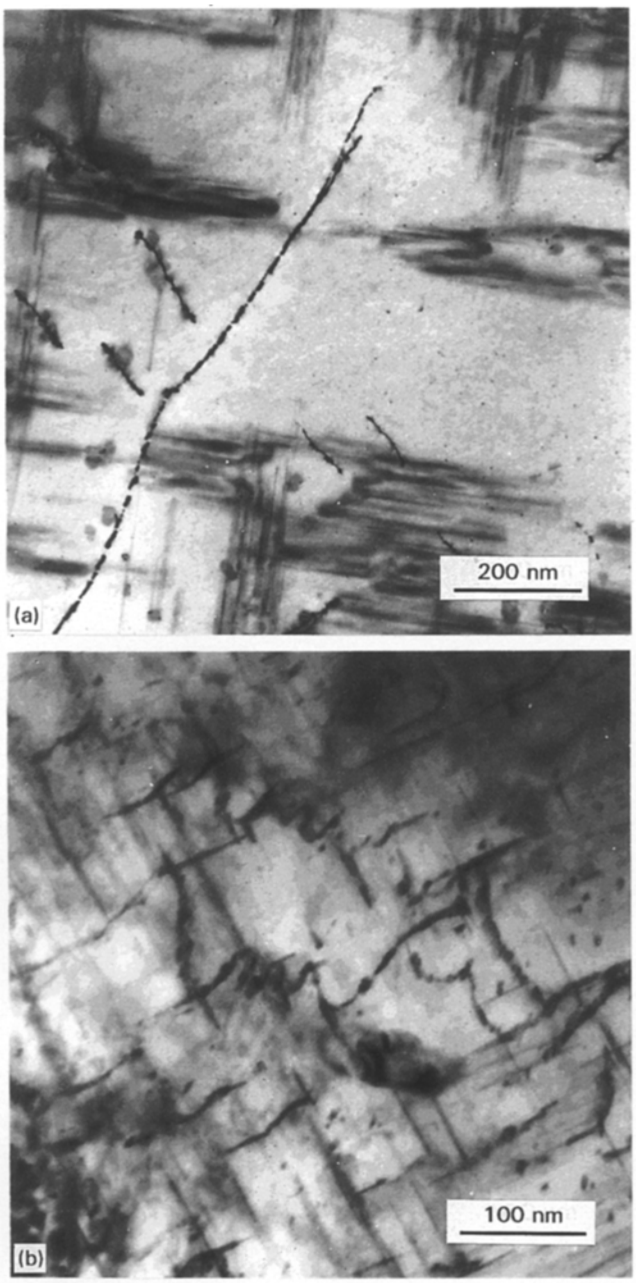

Figure 13 S′ precipitation after ageing for 144 h at 150 °C; (a) unreinforced Al–4%Cu–1.7%Mg–0.8%Mn alloy; (b) reinforced Al–4.45%Cu–1.5%Mg–0.13%Zr alloy.

Fig. 12(a–c). The addition of 0.12% zirconium was of no apparent benefit as far as refinement of the grain structure was concerned.

3.5. Effect of solution heat treatment and artificial ageing on fine microstructure

After solution heat treatment and ageing at 150 °C for 144 h, unreinforced Al–4%Cu–1.7%Mg–0.8%Mn sheet contained heterogeneous precipitation of S′(Al_2CuMg) phase on dislocation tangles, loops and helices formed as a result of vacancy condensation after quenching (Fig 13(a)). In contrast reinforced sheet contained considerably finer S′ precipitation in the peak aged condition, and appeared to be homogeneously and heterogeneously precipitated, with the latter decorating small dislocation helices (Fig 13(b)). Similar observations were made for the reinforced alloys containing 2% copper and 1% magnesium (alloys 1 and 2) although the dislocation helices appeared to be larger. Various dispersoids were evident in the Al–Cu–Mg alloys including Al–Cu–Mn-, Al–Cu–Si- and Al–Cu–Fe-rich particles probably originating from the casting of the pre-alloyed ingots.

In the naturally aged condition reinforced Al–4.35%Cu alloy sheet contained dislocation tangles

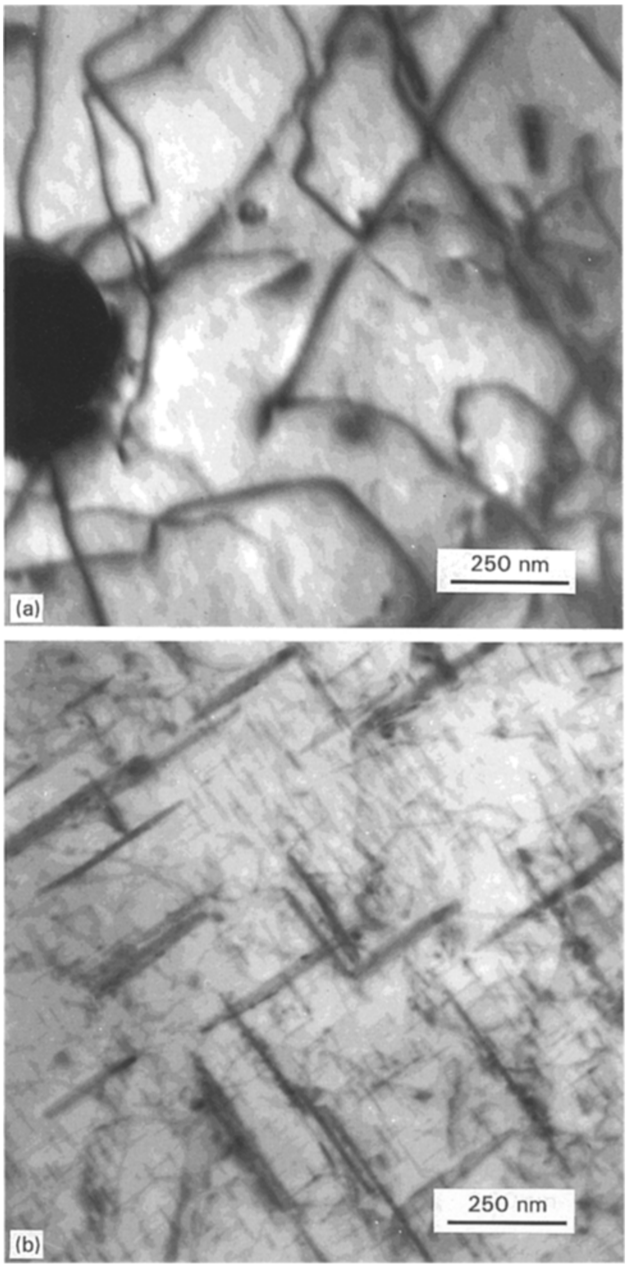

Figure 14 Fine microstructure of reinforced Al–4.35%Cu sheet after solution heat treatment for 30 min at 505°C and cold water quenching: (a) dislocation structure evident after natural ageing for 24 h; (b) Θ″ and Θ′ precipitation after ageing for 24 h at 150 °C.

and Al–Cu–Fe-rich dipersoids (Fig. 14(a)), but there was no evidence of precipitation. In contrast, after artificially ageing for 24 h at 150 °C widespread precipitation of θ″ and θ′ was evident (Fig 14(b)).

4. Discussion

It has been widely reported that in addition to their increased strength and modulus compared to unreinforced alloys, aluminium based MMCs exhibit an accelerated ageing response [1]. However, the tensile property data showed that whereas reinforcement of the Al–Cu–Mg alloys resulted in increased Young's moduli and strengths, accelerated ageing behaviour was not apparent. The hardness measurements on the reinforced alloys [4] nevertheless did indicate that accelerated ageing behaviour had occurred suggesting possibly that tensile properties are less sensitive to accelerated ageing effects. The accelerated ageing behaviour has often been explained in terms of enhanced nucleation and/or growth of precipitates in the heavily dislocated regions adjacent to the reinforcements [1]. The dislocations are generated during quenching after solution heat treatment owing to the thermal mismatch strains between the matrix and the reinforcing particles [5]. The Al–Cu–Mg alloys studied here containing nominally 4%Cu (alloys 3–5) fall within the 2124 alloy specification except that in alloy 5 zirconium was substituted for manganese. In this type of alloy the accelerated ageing behaviour has previously been explained by dislocation enhanced precipitation of S′ [1] although Hunt *et al.* [6] have shown that S′ precipitation was not readily apparent after natural ageing. After artificial ageing at 150 °C hardness [4] and tensile property measurents indicated that the onset of ageing for the unreinforced and reinforced alloys occurred after similar times and so ageing was not accelerated although the reinforcement did result in higher peak strengths and hardnesses. In the peak aged condition S′ precipitation was very fine and, in comparison to the unreinforced alloy, it was more homogeneously distributed. Assuming S′ phase was absent in naturally aged material as reported by Hunt *et al.* [6], S′ precipitation could have contributed to the observed increase in 0.2% proof stress (100 MPa) on artificial ageing to the peak aged condition but without contributing to accelerated ageing behaviour. Lowering the copper and magnesium content of the reinforced alloys to 2 and 1% respectively resulted in similar ageing behaviour although the strengths were lower, probably because of reduced S′ precipitation and reduced solution strengthening by copper and magnesium.

The manganese and zirconium additions seemed to have no effect on the tensile properties of the reinforced alloys although grain structure differences were observed with the higher manganese addition resulting in a finer grain structure.

The strengths of the Al–Cu-based MMCs were lower, and the ductilities higher, than for the reinforced Al–Cu–Mg alloys and no natural ageing occurred. This was not entirely unexpected since the natural ageing of Al–Cu alloys produced by ingot metallurgy is known to be slow [7]. The dislocation structures generated on quenching by the thermal mismatch strain between the matrix and the silicon carbide particles should result in higher strengths compared to the unreinforced alloys owing to dislocation strengthening. However Suresh *et al.* [8] have reported that enhancing the dislocation density in SiC reinforced Al–3.5%Cu MMC did not markedly contribute to strengthening. The artificial ageing response in Al–Cu MMCs, attributed to the co-precipitation of metastable θ″ and θ′ by Humphries *et al.* [9], increased the tensile properties considerably, although the strength levels obtained with the reinforced Al–4.35%Cu alloy were significantly lower than those for the reinforced Al–Cu–Mg alloys. A further strength improvement was obtained by increasing the copper content to 6% mainly as a result of increased solution strengthening and possibly increased precipitation. However the highest MMC strength obtained using this alloy was still 40–50 MPa lower than the values for the Al–Cu–Mg alloy MMCs can be attributed to a number of factors including the coarse grain structure, the absence of strengthening from magnesium and possibly a reduced dislocation density (since an Al–3.5%Cu-based MMC has been shown to have lower dislocation densities than a 2124 based MMC [8]). The suppression of natural ageing in Al–Cu-based reinforced alloys may be related to vacancy loss, either to the martix–particulate interfaces or to the dislocations generated during quenching. The lower strengths and absence of a natural ageing response in the Al–Cu alloy MMCs may be beneficial as far as fabrication processes are concerned, particularly as the strength can be improved later by artificial ageing. However the strength levels obtained are significantly lower than the target properties set by Peel *et al.* [10]. Further strength improvements may be possible by incorporating larger grain refining additions or by using manganese instead of zirconium, since this has been shown to be effective in the Al–Cu–Mg based MMCs. Cold rolling of the sheet [11] may also result in a finer grain structure and hence improved strength. Trace element additions such as indium, cadmium and tin may also prove beneficial since they have been shown to stimulate artificial ageing in cast Al–Cu alloys [12] without affecting the natural ageing behaviour and may also result in improved strength. However in the case of MMCs the influence of trace elements may be curtailed by the high dislocation density produced during quenching since they are reported to be ineffective in cast alloys which have been cold-worked prior to ageing [12].

The notch–yield ratio has been reported to give an indication of the relative toughness of high strength aluminium alloys [13]. In the present study the notch-yield ratios indicate that the Al–Cu based MMCs had a better toughness than the Al–Cu–Mg MMCs and also that the toughness in Al–Cu–Mg alloys was reduced by artificial ageing. Gregson [14] has shown that notch yield ratios of $\leqslant 1.0$ as obtained after artificial ageing to the peak strength condition, equate to plane stress fracture toughness values of < 20 MPa $M^{1/2}$, which are inadequate, and that acceptable toughness levels of ~ 50 MPa $m^{1/2}$ can be obtained with a notch yield ratio of 1.1 which corresponds to a 0.2% proof stress of ~ 400 MPa. The linear relationship obtained between the 0.2% proof stresses and notch-yield ratios (Fig. 10) suggests that toughness gains can be achieved by changes in matrix alloy composition but this may be associated with a reduction in strength.

5. Conclusions

1. The tensile properties' measurements showed no indications of accelerated ageing behaviour due to the presence of the reinforcement.

2. Reductions in the copper and magnesium levels in reinforced Al–Cu–Mg alloys from about 4 to 2% and 1.5 to 1% respectively had no effect on the rate of age hardening, but resulted in lower peak strength values.

3. There was no apparent effect of grain refining additions on the tensile properties of reinforced Al–Cu–Mg alloys. However manganese additions of 0.6 wt % were more effective in refining the grain structure of reinforced sheet than 0.12% zirconium additions.

4. The lower strength and absence of a natural ageing response in Al–Cu based MMCs may prove beneficial as far as fabrication processes are concerned, particularly as the strength can be improved later by artificial ageing.

5. Further improvements in the strength of reinforced sheet are desirable and may be achieved by larger grain refining additions, cold rolling or stretching prior to ageing. For the reinforced Al–Cu alloys trace element additions may also be beneficial.

6. Notched tensile tests revealed that Al–Cu MMCs were tougher than Al–Cu–Mg MMCs and that the toughness of Al–Cu–Mg MMCs was reduced by artificial ageing.

7. An improvement in the toughness of MMCs can be achieved by changes in matrix alloy composition but it may be associated with reduced strength.

Acknowledgements

The author gratefully acknowledges the sponsorship of this work by DCSA Ministry of Defence controlled by DR (S&C).

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Received 11 April 1994
and accepted 4 October 1994.