Microstructural evolution and ageing behaviour of the low Cu:Mg ratio Al–Cu–Mg alloys containing scandium and lithium

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Abstract

The microstructural evolution and ageing behaviour of Al–4.0Mg–1.5Cu–1.0Li–(0.2Sc)–0.12Zr alloys were investigated. The results indicate that small addition of Sc can enhance the ageing strengthening effect while the time taken to the ageing peak is prolonged. And the Sc-containing Al–4.0Mg–1.5Cu–1.0Li–0.2Sc–0.12Zr alloy exhibits a promising combination of strength and ductility. TEM observations show that the precipitation processes during ageing can be modified by the small addition of Sc.

Keywords: Aluminium alloys; Precipitation; Scandium; TEM

1. Introduction

Over the past several years there has been an increasing interest in investigating the beneficial effects achieved by adding Sc to Aluminium alloys. It has been demonstrated that small scandium additions can be favourable for a range of properties when added to Al–Mg, Al–Zn–Mg and Al–Li type of alloys [1–7]. The new emerging of Al–Mg–Sc series of alloys, such as 01570 (Al–6Mg–0.3Sc), is attributed to the high performance achieved by small additions of Sc in Al–Mg alloys [8,9]. The addition of trace scandium, especially the addition of Sc and Zr simultaneously to Al–Zn–Mg alloy has led to developing a new Al–Zn–Mg–Sc series alloys, such as 01970 (Al–5.2Zn–2.0Mg–0.2Sc–0.1Zr). As for Al–Li alloys, several new alloys microalloyed by Sc have been developed and partly used as new light materials for next generation applications. And the successful development of 1421 (Al–5Mg–2Li–0.2Sc–0.1Zr) [10] and 1460 (Al–3Cu–2Li–0.2Sc–0.1Zr) [11,12] alloys are typical examples. Despite their commercial success and the fact that considerable research has already been done, the understanding of their underlying precipitation processes remains incomplete. Besides, little work has been reported in Al–Mg–Cu–Li alloy containing trace scandium up to now. In this paper the effect of small addition of Sc on the microstructural evolution and properties in the Al–4.0Mg–1.5Cu–1.0Li–0.12Zr alloys have been investigated.

2. Experimental procedure

The alloys were prepared by melting and casting under argon atmosphere, and their normal compositions are provided in Table 1. The ingots were homogenized, scalped, then hot rolled and finally fabricated to 2 mm sheets by cold rolling. The specimens were solution treated and quenched into cold water, then aged at 200 and 170 °C respectively. Tensile specimens were machined from the sheets in the longitudinal directions, and the tensile tests were carried out at room temperature on Instron 8019. Foils for transmission electron microscopy (TEM) were electro-polished using a 33% nitric acid–67% methanol solution at around −40 °C. TEM was performed on a Philips CM12 microscope operating at 120 kV with a LaB6 filament for conventional bright-field (BF) and dark-field (DF) imaging, selected area electron diffraction (SAED) and energy-dispersive X-ray (EDX) microanalysis.

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3. Results

3.1. Tensile properties

The room-temperature tensile properties/ageing time curves at 200 °C for alloy A and B are presented in Fig. 1. With increasing ageing time, the tensile strength of Sc-free Al–4.0Mg–1.5Cu–1.0Li–0.12Zr alloy increases to the peak value and then decreases, well correlative with the conventional age-strengthening behaviour. As for the Sc-containing alloy B, there is no remarkable increment observed for the tensile strength with respect to ageing time, while the strength value keeps fairly high level from the early ageing stage. As to the percentage elongation, it decreases gradually with increasing ageing time in Sc-containing B alloy, whereas a remarkable decrease is observed after peak ageing in alloy A.

Fig. 2 shows the tensile properties of the two experimental alloys aged at 170 °C with respect to ageing time. It can be seen that the trend of the ageing curves of both alloys aged at 170 °C is very similar to that at 200 °C. It is apparent that a remarkable higher strength can be achieved by the addition of Sc, while the time taken to peak strength for the Sc-containing alloy is longer than Sc-free alloy A. Compared with the strength when ageing at 200 °C, the strength at 170 °C is much higher, and the time taken to peak strength is increased when aged at lower temperature.

3.2. Microstructure

Fig. 3 shows the transmission electron micrographs of the Al–4.0Mg–1.5Cu–1.0Li–(0.2Sc)–0.12Zr alloys at early ageing stage (1 h) at 170 °C. It can be seen that the microstructure is dominated by some particles and dislocation loops in the Sc-free Al–4.0Mg–1.5Cu–1.0Li–0.12Zr alloy. The results from the EDX analyses indicate that Zr is present in the particles, which suggests that Al3Zr particles precipitate in the alloy A aged at 170 °C for 1 h. However, in the Sc-containing alloy B there are $\delta'$ (Al3Li) and spherical compound particles precipitated. EDX analyses show that there appears to be a significant amount of Sc and Zr in the inner region of the compound particles, yet none in the outer shell. It is concluded that the outer shell probably consists of $\delta'$ precipitate similar to that formed in the matrix. Thus the compound particles is Al3(Sc,Zr)/Al3Li.

BF and DF TEM images of the peak aged alloy A and alloy B during ageing at 170 °C were recorded close to
the \( \langle 011 \rangle _{s} \) zone axis and examples are provided in Fig. 4. There are sparsely distribution of \( \delta' \) precipitates and heterogeneously distribution of \( \text{Al}_3\text{Zr} \) particles together with \( S \) precipitates along dislocation in the Sc-free alloy A, while the microstructure of the Al–4.0Mg–1.5Cu–1.0Li–0.2Sc–0.12Zr alloy is dominated by the homogeneously distribution of \( \delta' \) and spherical \( \text{Al}_3(\text{Sc},\text{Zr})/\text{Al}_3\text{Li} \) compound particles. Compared to the dimension of \( \delta' \) and spherical \( \text{Al}_3(\text{Sc},\text{Zr})/\text{Al}_3\text{Li} \) compound particles precipitated in the Al–4.0Mg–1.5Cu–1.0Li–0.2Sc–0.12Zr alloy at early ageing stage (1 h), negligible growth was observed, which suggests that those precipitates are resistant to coarsening during ageing.

The transmission electron micrographs of the peak aged Al–4.0Mg–1.5Cu–1.0Li–(0.2Sc)–0.12Zr alloys during ageing at 200 °C are presented in Figs. 5 and 6. It can be seen that there are rod-shape precipitates oriented along \( \langle 100 \rangle_s \) direction in both Sc-free and Sc-containing alloys, and their corresponding SAED pattern shows sharp streaking through \( \{100\}_s \) (Fig. 6(a)), these results are consistent with GPB (Guinier–Preston–Bagaratsky) zone. In summary, besides the precipitates present in the alloy A aged at 170 °C, there are homogeneously distribution of GPB zone precipitated in the Sc-free Al–4.0Mg–1.5Cu–1.0Li–0.12Zr alloy aged at 200 °C for 8 h.

Fig. 3. Transmission electron microscopy results from (a) Al–4Mg–1.5Cu–1.0Li–0.12Zr alloy and (b and c) Al–4Mg–1.5Cu–1.0Li–0.2Sc–0.12Zr alloy aged at 170 °C for 1 h. The electron beam is parallel to \( \langle 001 \rangle_s \).

Fig. 4. Transmission electron micrographs of the Al–4Mg–1.5Cu–1.0Li–(0.2Sc)–0.12Zr alloys aged at 170 °C for: (a) 55 h of the Al–4Mg–1.5Cu–1.0Li–0.12Zr alloy; (b) 100 h of the Al–4Mg–1.5Cu–1.0Li–0.2Sc–0.12Zr alloy; (c) 55 h of the Al–4Mg–1.5Cu–1.0Li–0.12Zr alloy; (d) 100 h of the Al–4Mg–1.5Cu–1.0Li–0.2Sc–0.12Zr alloy; The electron beam is parallel to \( \langle 011 \rangle_s \).
Fig. 5. Transmission electron micrographs of the Al–4Mg–1.5Cu–1.0Li–0.12Zr alloy aged at 200 °C for 8 h: (a) BF image; (b) DF image; (c) corresponding SAED pattern.

Fig. 6. Transmission electron micrographs of the Al–4Mg–1.5Cu–1.0Li–0.12Sc–0.12Zr alloys aged at 200 °C for 10 h: (a) BF image; (b) DF image; (c) corresponding SAED pattern; (d) DF image.
TEM investigation indicated that the microstructure of the Al–4.0Mg–1.5Cu–1.0Li–0.2Sc–0.12Zr aged at 200 °C for 10 h is dominated by homogeneously distribution of fine δ′ and spherical Al13(Sc,Zr)/Al3Li compound particles together with GPB and S precipitates (Fig. 6).

4. Discussion

Investigations carried out earlier [13] have shown that GPB zones can be formed in the Al–Cu–Mg alloys aged at 200 °C with the Cu:Mg ratio decreasing systematically from 35 to 2.6. And there are two striking features of GPB zones. The first is their rapid rate of formation, compared to Guinier–Preston (GP) I zones, while the second concerns the wide range of thermal stability exhibited. The present study shows that GPB zone present in both Sc-free alloy A and Sc-containing alloy B aged at 200 °C.

As we know, the ageing strengthening is derived from the nucleation and growth of coherent metastable δ′ precipitate(Al3Li,LI2 structure) in Al–Li alloys. δ phase can be precipitated during quenching after solution treated, and grow up during ageing process. The nucleation of δ′ on the Al3(Sc,Zr) interface concomitant with nucleation of δ′ in the matrix can be considered as a physical interaction between these precipitates [14]. Nucleation on pre-existing second phase particles has been seen in many materials, and it can be explained by considering the surface area of the precipitates formed. The interfacial energy resulting from precipitation at a pre-existing interface particles can be reduced compared to homogeneous nucleation in the matrix. As the interfacial δ coalesces to form a complete shell, the compound precipitate will act as though it was wholly δ′ and thus coarsen like a δ′ particle of increased radius. It’s found that the coarsening of δ′ obeys Lifshitz–Slyozov and Wagner theories of particle coarsening [15,16]. Here the average δ′ particle radius r0 at time t0 is related to the average δ′ particle radius rt at time t by the coarsening rate constant, k, through:

\[ k = \frac{8γC_mD_j^2T}{9RT} \]

where γ is the interfacial energy between δ′ and matrix, Cm the equilibrium solubility of Li, D the diffusion coefficient for Li atoms, Vm the molar volume of δ′ phase, and T is ageing temperature. Since R and T are constants, and only negligible change is expected for Vm and γ, the diffusion coefficient D is an essential factor that can largely change the value of the coarsening rate constant K in the above equation. Recent work by Miura et al. [17], the binding energy of Sc with vacancy was determined to be 0.35 eV, which is higher than those for other solutes, such as Li(0.26 eV) and Mg(0.25 eV) [18]. Consequently the diffusion rate of Li atoms will be decreased by the addition of Sc. The low growth rate of δ′ and spherical Al13(Sc,Zr)/Al3Li compound particles in the Sc-containing Al–4.0Mg–1.5Cu–1.0Li–0.12Zr alloy is mainly due to the decrease in diffusivity, D. In summary, the present study results show that the δ′ and spherical Al13(Sc,Zr)/Al3Li compound particles are resistant to coarsening during ageing, and the slow growth rate of δ′ and spherical Al13(Sc,Zr)/Al3Li compound particles may be because scandium atoms retard migration of Lithium atoms in the matrix due to high binding energy between scandium atoms and vacancies.

5. Conclusion

1. Small addition of Sc can enhance the ageing strengthening effect of the Al–4.0Mg–1.5Cu–1.0Li–0.12Zr alloy, while the time taken to the ageing peak is prolonged by small addition of Sc.

2. Significant improvements in mechanical properties are obtained by small addition of Scandium to the Al–4.0Mg–1.5Cu–1.0Li–0.12Zr alloy.

3. The precipitation processes during ageing can be modified by the small addition of Sc, what is of special significance is the observation that the microstructure of the aged Sc-containing alloy has fine and uniform dispersions of δ′ and spherical Al13(Sc,Zr)/Al3Li compound particles.

References