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Microstructural characteristics and mechanical properties of Al–2.5 wt.% Li–1.2 wt.% Cu–xMg alloys

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Abstract

Microhardness of heat-treated Al–2.5 wt.% Li–1.2 wt.% Cu–0.12 wt.% Zr–xMg alloys was observed to increase with increasing Mg content mainly due to increased solid solution strengthening, reduced grain size and increased amount of fine quasi-crystalline $Al_6Cu(Li, Mg)_3$ precipitates. Although coarse quasi-crystalline $Al_6Cu(Li, Mg)_3$ phases were found in all investigated alloys, the quasi-crystalline phase as a fine precipitate could be only observed in the alloys containing high Mg contents. The tensile strength was significantly increased in the alloys with high Mg contents after a heat treatment, owing to the enhanced precipitations of fine Al_3Li and Al_3Zr phases.

Keywords: Metals and alloys; Phase transitions; Transmission electron microscopy

1. Introduction

Al–Li alloys have attracted much attention as promising structural materials in the aircraft and aerospace industries because of their low density and high stiffness. However the practical application area is still limited partly because only a few commercial Al–Li alloys are available, and the development of new Al–Li base alloys with more diverse properties is expected to significantly expand the application area.

In 8090 (Al–Li–Cu–Mg–Zr) alloys δ' (Al_3Li) has been known as a main precipitate responsible for the high strength [1], but other phases such as S′ (Al_2CuMg), T_1 (Al_2CuLi) or T_2 (Al_6CuLi_3) can give an additional contribution to the overall strength [2,3]. Extra Mg additions to 8090-type Al–Li–Cu–Mg–Zr alloys would result in more significant weight savings and likely an increased strength because of the low density and high solubility of magnesium. It is expected that the Mg addition lowers the solubilities of Li and other alloying elements in Al matrix, thereby facilitating the precipitations of some precipitates. Mukhopadhyay et al. [4] have reported that the relative stability of T_2 (Al_6CuLi_3) and other Al–Li–Cu–Mg phase (C phase) was affected by Cu:Mg and Cu:Li ratio. Although a number of researches have been conducted to investigate various precipitates formed in Al–Li–Cu–Mg alloys [5–7], quite a few questions still remain as to the evolution of the precipitates and the resultant mechanical behaviors. Especially, the effect of extra-additions of Mg to 8090 base alloys on the microstructure and mechanical properties has not been clearly understood.

2. Experimental

Al–2.5% Li–1.2% Cu–0.12% Zr (wt.%) alloys containing different Mg contents (1, 1.5, 2 and 2.5%, respectively) were prepared by melting 99.9% pure metals and master alloys in a vacuum induction furnace under inert gas atmosphere. Plate shaped castings (60 mm × 100 mm × 24 mm) were fabricated in a metallic mold and used for subsequent rolling and heat treatment processes. The rolling was carried out to reduce the thickness of plate specimens from 5 to 2 mm with the thickness reduction ratio of 20%. The solid solution treatment was conducted at 500 °C for 3 h, followed by water quenching.

The microhardness change of investigated alloys was measured after various aging treatment, and tensile tests were carried out, according to ASTM B 557M, on the specimens that were heat-treated under the optimum condition. Microstructural characterization was performed using optical microscope, scanning electron microscope (SEM) equipped with energy dispersive X-ray spectrometer (EDS), and transmission electron microscope (TEM) equipped with EDS.

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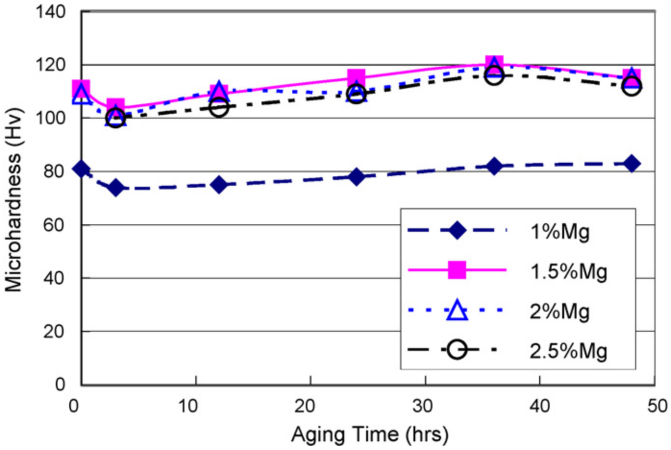

Fig. 1. Microhardness variations of Al–Li–Cu–Mg alloys during the aging at 160 °C.

3. Results and discussion

The microhardness of investigated alloys was generally observed to increase during the aging at 160 °C, as shown in Fig. 1. This hardness increase is believed related to the precipitation of fine precipitates such as δ' (Al_3Li). It is also worthwhile to note a reversion of hardness occurred at the beginning of the aging, and the dissolutions of GP zones and some of δ' (Al_3Li) phases are suggested to be the reason for this phenomenon [2,5]. Another interesting feature shown here is that the initial hardness of 1% Mg-added alloy is significantly lower than those of other alloys containing higher Mg contents. Increased solid solution strengthening and grain refining (Fig. 2) by the increased Mg content should be responsible for that, however, those effects alone hardly explain why the alloys with 1.5–2.5% Mg possess the similar initial hardness. Another reason may be the existence of relatively fine quasi-crystalline phase. A typical TEM micrograph with a diffraction pattern in Fig. 3 indicates the quasi-crystalline phase with a quasi-crystalline structure and an average diameter of 200 nm. This phase is most likely to be T_2-phase, $Al_6Cu(Li, Mg)_3$, [4,8] and was observed only in the high Mg alloys.

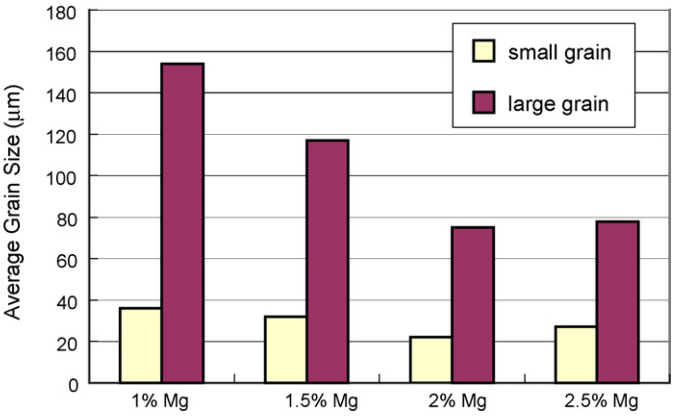

Fig. 2. Average grain size of small and large grain groups formed in Al–Li–Cu–Mg alloys.

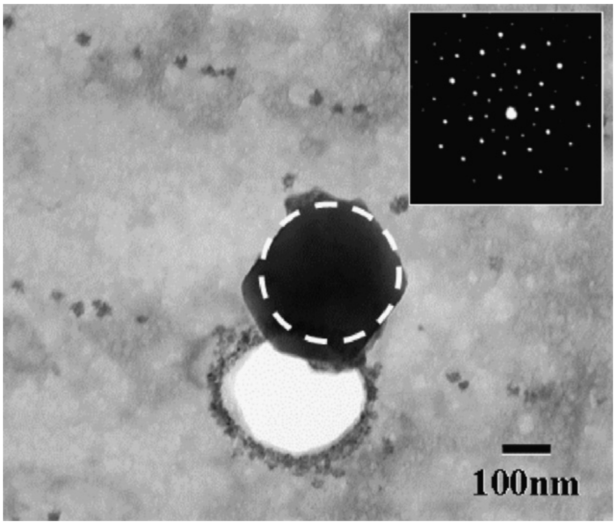

Fig. 3. Typical TEM micrograph and the diffraction pattern (five-fold symmetry) obtained from a fine quasi-crystalline phase (dashed circle) in high Mg alloys.

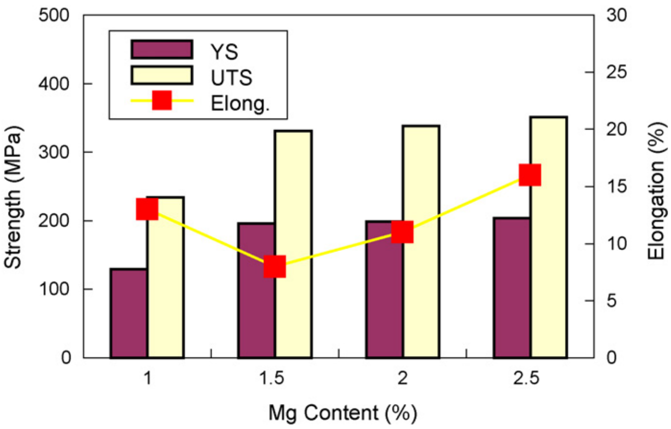

Fig. 4. Tensile properties of peak-aged Al–Li–Cu–Mg alloys.

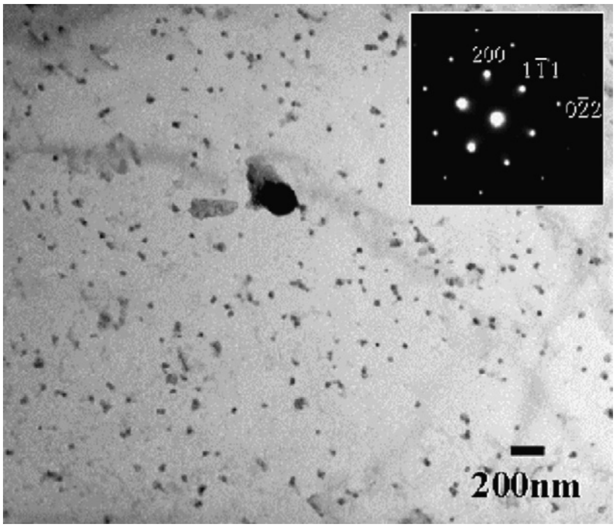

Fig. 5. TEM micrograph of fine precipitates observed in peak-aged Al–Li–Cu–Mg alloys (SADP inset for matrix, [1 1 0]-direction).

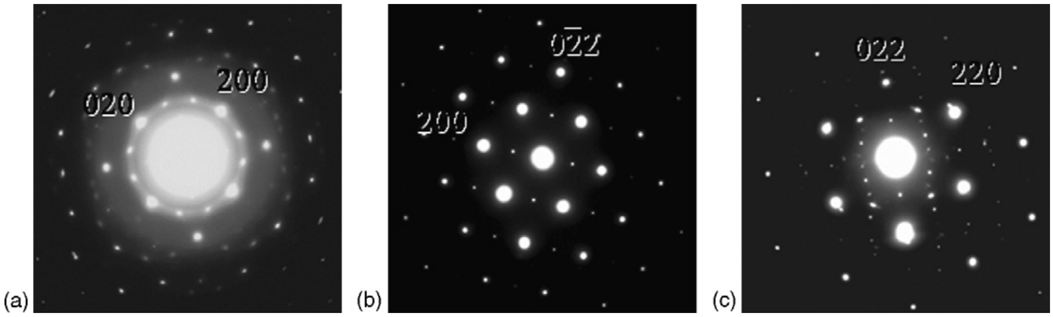

Fig. 6. Diffraction patterns for matrix, also showing the spots for the fine precipitates: (a) [1 0 0]-direction; (b) [1 1 0]-direction; (c) [1 1 1].

The tensile properties of peak-aged alloys are presented in Fig. 4. The tensile strength of the high Mg alloys appears much higher than 1% Mg-added alloy, just like the case of microhardness. In 8090 base alloys the largest contribution to strength comes from the precipitation hardening by δ' (Al_3Li) and other fine phases. Fig. 5 shows a typical micrograph of the fine precipitates. TEM observations revealed that comparatively similar characteristics (the amount and size) of the precipitates were found in the investigated alloys, regardless of Mg content, except for 1% Mg. In 1% Mg-added alloy the amount of fine precipitates was lower and they were less uniformly distributed. This may be due to the higher solubilities of Li and Zr occurred by the lower Mg content, resulting in the lower tendency of precipitation. Fig. 6 indicates the diffraction patterns of matrix that also include the spots for the fine precipitates, δ' (Al_3Li) and Al_3Zr. The superlattice spots in (b) evidence the presence of δ' (Al_3Li) phase, while the extra spots at $\{220\}/2$ in (c) suggest the existence of Al_3Zr phase [5].

All the investigated alloys showed two different coarse Al–Li–Cu–Mg phases (one includes much higher Cu content than that of the other). It is postulated that the phase containing higher Cu is T_2–$Al_6Cu(Li, Mg)_3$ and the other is C-phase, based on EDS analysis results [4]. Morphological difference between two phases was not clearly shown, however, T_2-phase often appeared somewhat coarser compared to C-phase. Although the relative stability of the two phases has been known to depend on the relative Cu. Li, and Mg contents [4], any apparent trends could not be detected in this research. It seems interesting that the ductility of alloys is continuously increased as Mg content in the range between 1.5 and 2.5%. One of the reasons for this may be the reduced grain size and increased uniformity (reduced size difference between small and large grains, the grains were selected at random and classified into two groups by comparing their grain size with the average), as shown in Fig. 2. Solute Mg atoms can act as obstacles to the migration of boundaries and effectively prevent recrystallization and coarsening during the thermo-mechanical process.

4. Summary and conclusions

Coarse quasi-crystalline $Al_6Cu(Li, Mg)_3$ phases were found with Al–Li–Cu–Mg phases (C-phase) in all the investigated alloys, however the quasi-crystalline phase as a fine precipitate type could be only observed in the alloys containing high Mg contents. The microhardness and tensile strength were significantly higher in the high Mg alloys, compared to low Mg alloy, because of the increased solid solution strengthening by Mg, reduced grain size, and enhanced precipitation of fine phases such as Al_3Li, Al_3Zr and $Al_6Cu(Li, Mg)_3$.

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