The strengthening mechanism of deformed and aged Al-Cu-Mg alloy

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The microstructure and mechanical properties of deformed and subsequently aged Al-Cu-Mg alloy were investigated by employing transmission electron microscopy and hardness test. It was found that when the deformed alloy with 5%-strain was aged to peak temper, the strengthening mechanism was ascribed to the small dispersed S phases, which was different from the GPB zone as strengthening particle in conventional ageing. Besides, the main strengthening mechanism in the initial ageing stage was work hardening effect due to quantities of dislocation.

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1. Introduction

Aluminum alloy has been widely used on aircraft and spacecraft industries after certain heat treatment. In order to obtain excellent mechanical properties, improve the efficiency and save energy, thermo-mechanical treatment is employed on this alloy. The mechanical properties of the alloy are strongly dependent on the thermo-mechanical process and the strengthening mechanism. During thermo-mechanical treatment process, dislocation hardening and precipitate hardening are two important mechanisms to strengthen the alloy. So it is meaningful to understand the relationship between the microstructure and mechanical properties for better utility of thermo-mechanical treatment. Some authors have investigated the influence of the precipitation on the work hardening behaviors, and propose models to enables the overall mechanical response for a variety of ageing conditions [1-6]. Besides, it is also found that the cold work prior to elevated-temperature ageing can reduce the size of the precipitates in Al-Cu-Mg, Al-Cu-Li-Mg-Ag and Al-Cu-Li alloys [7,8]. It has been generally regarded that the GPB zones are the predominant strengthening precipitates in Al-Cu-Mg alloys [9-11]. However after the cold work, which is the strengthening precipitate is not thoroughly clear. The present work is to investigate the strengthening mechanism of commercial Al-Cu-Mg aluminum alloy.

2. Experimental

The chemical compositions of commercial Al-Cu-Mg alloy used in this study is listed in Table 1. Specimens were solution treated at 495℃ for 1h and quenched into water at room-temperature. And then the specimens were tensile stretched 5 pct immediately upon quenching. The deformation was performed at a crosshead displacement of 1mm/min with an Instron machine. The nondeformed and 5 pct deformed specimens were artificially aged at 170 °C in oil bath furnace to peak age. Vickers hardness measurement was employed with a load of 5kg and dwell time of 20s. Thin foils for transmission electron microscopy (TEM) were electro-polished in a twin-jet Tenupol by a 33% nitric acid solution in methanol at -25˚C. FEI CM20 and JEOL 2100F TEM were operated at 200kV.

Table 1. Chemical composition of commercial Al-Cu-Mg alloy (wt%).

Cu	Mg	Mn	Cr	Ni	Zn	Ti	Si	Fe	Al
4.2	1.5	0.62	0.01	0.01	0.02	0.01	0.03	0.11	balance

3. Results

The effect of cold work on the ageing hardness curves at 170℃ of the alloy with different strains of 0 and 5% is shown in Fig. 1. After the specimen was deformed for 5% and aged to peak` temper, the peak hardness is about 170 VHN larger than that without deformation 142 VHN. Meanwhile, the time to reach the peak temper is 36h, which is greatly shorter than the conventionally aged specimen (72h). And the deformed specimen is harder than the undeformed specimen no matter underage or overage. So that it can be clearly seen that the deformation can increase the hardness and reduce the time to peak temper.

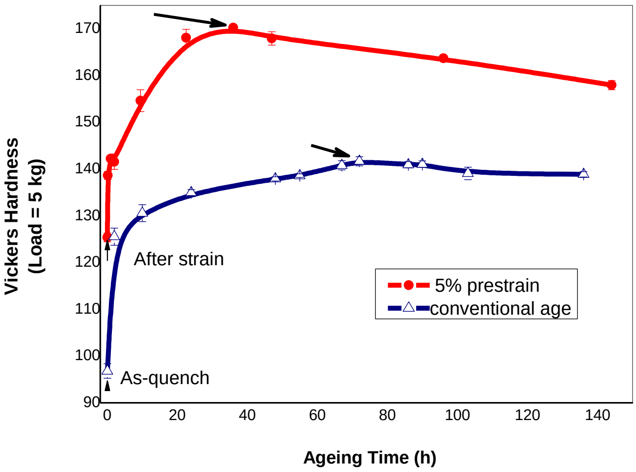

Fig. 1. Ageing hardness curves of alloys.

After solution treated and followed by 0% or 5% pre-deformation, the microstructure of the alloys was observed by TEM as shown in Fig. 2. The TEM micrographs were recorded near the <001> incident beam and corresponding selected area diffraction (SAD) patterns were recorded parallel to $<001>_{\alpha}$ zone axes. It can be seen that except the intermetallic phase (black rod shaped phase) there is no detectable precipitate in as-quenched alloy in Fig. 2 (a). Fig. 2(b) is the dark field image of the 5%-deformed alloy. A number of dislocations can be observed in the alloy. The corresponding SAD patterns also reveal that no precipitates can be detected in the pre-deformed alloy with 5% pre-strain.

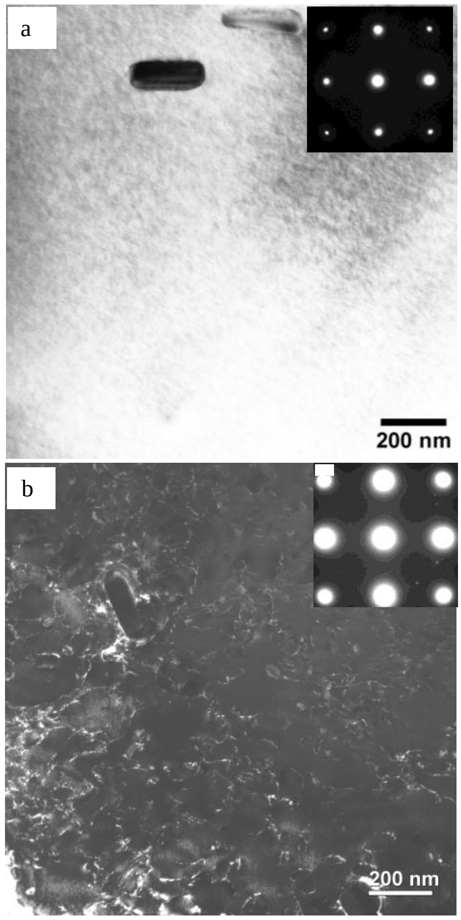

Fig. 2. $<001>_{\alpha}$ TEM micrographs (a) 0%, (b) 5%.

Fig. 3 shows the <001>□bright field (BF) TEM micrographs and SAD patterns of the peak-aged alloy with or without deformation. Fig. 3(a) and Fig. 3(b) are the microstructures of the peak-aged alloy without deformation. Fig. 3(c) is the deformed alloy with 5%. There are two kinds of phases in undeformed alloy: S phases (large laths) as shown in Fig. 3(a) and GPB zones (tiny particles) as shown in the magnified image Fig. 3(b). The relationship of S phase and the matrix is $\{100\}_S//\{012\}_{Al}$ and S phases are present as laths along <001>. So the size of S phase can be easily estimated by the length of the lath. Thus it’s obviously to observe that the size of S phase in Fig. 3(a) is larger than that in Fig. 3(c). And the distribution of the S phases is more uniform in Fig. 3(c).

Generally, tiny particles GPB zones are considered as the predominant contributions to strength. It can be seen GPB zones are homogeneously distributed in the matrix in Fig. 3(b). And the existence of GPB zones can also be proved by SAD pattern in Fig. 3(b). Dense, uniform and tiny GPB zones distribute in the matrix in the peak-aged alloy without deformation. However, it is hardly to observe GPB zones in Fig. 3(c). This indicates that the precipitate hardening effect is not due to GPB zones but S phase.

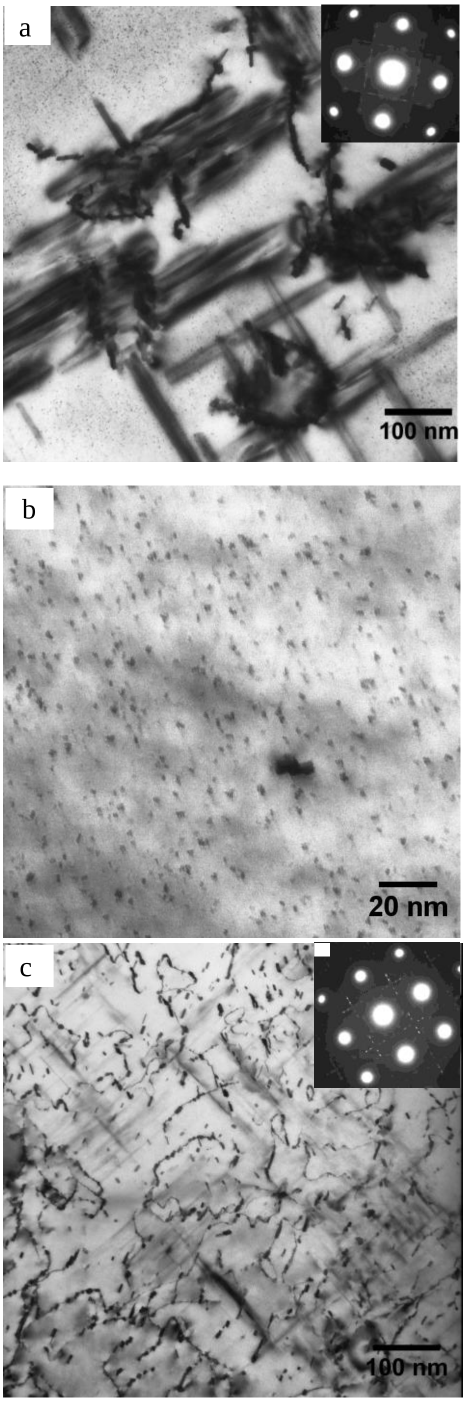

Fig. 3. <001>$_\alpha$ BF TEM micrographs (a) (b) 0% (c) 5% of peak-aged samples.

In order to further prove that the precipitate in the peak-aged specimen is S phase not GPB zones, high resolution transmission electron microscopy (HRTEM) was employed as shown in Fig. 4. Again, Fig. 4(a) proves that there is only one kind of precipitate. Fig. 4(b) is the magnified particles in Fig. 4(a). As analyzed in Fig. 4(b), the atomic parameters of this particle are 0.72nm and 0.46nm, and the orientation relationship with matrix is $\{100\}_S$ // $\{012\}_{Al}$, which match with the S phase perfectly. Fig. 5(a) is the FFT image of Fig. 4(b). It also matches very well with the simulated SAD pattern of one variant of S phases as shown in Fig. 5(b). Thus, it can be concluded that the precipitate in the peak-aged alloy with 5 pct pre-strain is S phase. Meanwhile, it can be deduced that the strengthening precipitate is also S phase.

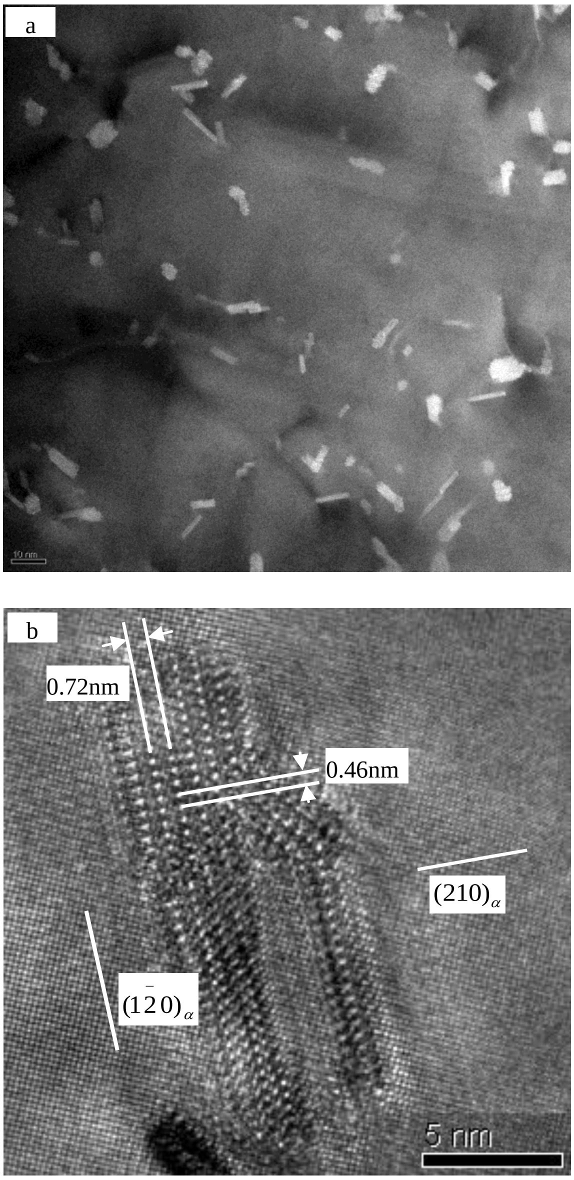

Fig. 4. HRTEM image at low magnification (a) and HRTEM image of S phase (b).

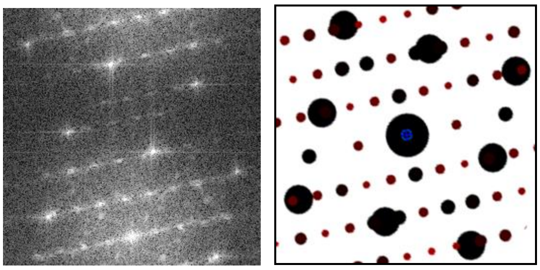

Fig. 5. FFT of Fig.4(b) and simulated diffraction pattern of S phase Big circles-Al reflections, small circles-S phase.

4. Discussion

The main strengthening mechanisms of alloy dependent on the thermo-mechanical process are dislocation hardening and precipitate hardening. The contribution of hardness during this process can be mainly owed to the work hardening effect and precipitate hardening. This can be expressed by equation (1).

$$\sigma = \sigma_0 + \sigma_{ss} + \sigma_{\perp} + \sigma_{ppt} \quad (1)$$

σ_0 is the so called intrinsic strengthening due to aluminium lattice resistance, σ_{ss} is the contribution of solid solution. These two items can be neglected here compared with the other two items. $\sigma_{\perp}$ and σ_{ppt} are the dislocation strengthening and precipitate strengthening respectively. $\sigma_{\perp}$ varies dependent on the average density of obstacle. After given a deformation about 5 pct on the alloy, the hardness increases greatly compared with the quenched alloy without deformation. As seen in Fig. 2(a), there is no dislocation or precipitate in the quenched alloy, so that the hardness value is quite low. But Fig. 2(b) presents lots of dislocations. High density of dislocation is generated and dislocation strengthening $\sigma_{\perp}$ plays a large role to the strength of the alloy. At this stage, the alloy is not aged and the precipitates haven't emerged. Thus, the increment of hardness in the initial stage can be ascribed to the work hardening effect.

S phase will nucleate preferentially on dislocations. With the ageing process proceeding, S phase begins to grow. The precipitate is still tiny in the early ageing stage. Therefore, the precipitate is easily to be cut through which can be explained by Orowan mechanism. Prolonging the ageing time, S phase grows large enough to hinder dislocations. At the same time the density of dislocation decreases due to the growth of S phase consuming dislocations. As a result the precipitate strengthening effect increases and dislocation strengthening decreases. Consuming of solute atoms leads to the reduction of concentration of solute atoms in the matrix so that the length of S phases is small, which is consistent with other authors [12,13].

Cu:Mg ratio is 2.8 which is in the region of α+S phase, and the main precipitates are GPB zone and S phase. Generally the peak strengthening precipitate is considered to be dispersed tiny GPB zones. S phases are coarse and play a relative small role on the strength of the alloy. After the alloy is deformed preferentially formed S phases cause low concentration of vacancies, which results in GPB zones are not easy to generate. Even if some GPB zones produce in the later ageing process, they will gradually disappear due to Oswald ripening. Then S phases are the main precipitates in the alloy. As seen in Fig. 4, the uniformly dispersed S phases are the dominant strengthening precipitates in the deformed and aged Al-Cu-Mg alloy.

5. Conclusions

A study of the strengthening mechanism of deformed Al-Cu-Mg alloy was carried out by applying 5% deformation before age treatment. The conclusions can be drawn as followings.

(1) The cold work before age treatment can increase the hardness of the alloy due to the work hardening effect. Meanwhile, it can shorten the time to reach peak age.

(2) The peak hardness is mainly strengthened by dispersed small S phase in the 5% deformation Al-Cu-Mg alloy not GPB zones as in conventional ageing.

Acknowledgements

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