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Study of work hardening of quenched and naturally aged Al–Mg and Al–Cu alloys

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ABSTRACT

A comparative study on the work hardening of Al–Mg and Al–Cu alloys was carried out using a Kocks–Mecking–Estrin type analysis of stress–strain curves obtained in tension tests at constant loading rate. As a result of the analysis, dependencies of forest dislocation storage and dynamic recovery rates on the Mg and Cu concentration have been derived. The work hardening behavior and the microstructure formation in the Al–Mg and Al–Cu alloys were shown to be similar despite the opposite effects of Cu and Mg on stacking fault energy as well as the differences in solute atom size and friction stress. The influence of alloying on the work hardening peculiarities and the dislocation substructure evolution was discussed in connection with the effects of solute–dislocation interaction.

1. Introduction

Dissolved atoms and their clusters can make a considerable contribution to the resistance of aluminium alloys to plastic deformation but are known to produce minor effect on the alloys work hardening [1]. Nevertheless, the latter effect exists, and quantitative differences in the work hardening behavior of various alloys take place [2–5]. Work hardening is associated with the evolution of dislocation substructure, which, in turn, is determined by two alternative processes [6,7]. One is athermal storage of forest dislocations resulting from the interaction of mobile dislocations with various obstacles. The other is dynamic recovery, which determines a reduction of the dislocation density as the result of thermally activated rearrangements and annihilation. The effect of dissolved atoms of an alloying element on the work hardening is due to their effect on these two processes.

The microstructure evolution and strain hardening have been investigated extensively in the case of Al–Mg alloys [2,8–13]. It was shown that the addition of Mg inhibits cell structure formation [8,9]. That phenomenon is often supposed (e.g., [12]) to be associated to the fact that Mg reduces the stacking fault energy (SFE) [14,15]. Actually, such an effect of SFE reduction is established for many metals and alloys. At the same time, in aluminium alloys, unlike for many other materials, the splitting of dislocations is very feebly marked [7]. Moreover, the early data on the SFE obtained from mechanical testing (e.g., Ref. [14]) are considered as unreliable because of being based on an inaccurate cross slip model [7], while the estimations based on the annealing rate of faulted dislocation loops did not show that Mg addition lowers significantly the SFE of Al [16].

In this connection the case of Al–Cu alloys is of interest since copper influences SFE of aluminium in the opposite direction as compared to magnesium [15]. Another peculiarity of Al–Cu alloys is a considerable hardening observed at room temperature immediately after quenching [3,17], whereas natural ageing has virtually no effect on the strength of Al–Mg alloys [3]. Unfortunately, the data concerning Al–Cu alloys are fragmentary and are related mainly to large strains [18,19]. Earlier experimental data and models concerning work hardening behavior of Al–Cu and Al–Mg alloys have been surveyed by Lukac [11]. In a recent EBSD study [20] the orientation fragmentation in Al–Cu and Al–Mg alloys was examined after rolling. It was shown that qualitatively similar structures develop in these alloys; at the same time adding Cu hampers fragmentation, as compared with pure aluminium, to a greater extent than Mg.

In this paper, a comparative study of Al–Mg and Al–Cu alloys is presented. The aim is to examine the alloying effect on the strain hardening and on the microstructure evolution in aluminium alloys as function of Mg and Cu concentrations.

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Table 1
Chemical compositions of the alloys (at.%).

Alloy	Mg	Cu	Al
Al–1.9% Cu	0.001	1.92	Balance
Al–1.5% Cu	0.001	1.45	Balance
Al–0.6% Cu	0.001	0.57	Balance
Al–5.6% Mg	5.56	0.001	Balance
Al–2.5% Mg	2.48	0.001	Balance
Al–0.8% Mg	0.81	0.001	Balance
Al	0.001	0.001	Balance

2. Experimental

Al–Cu and Al–Mg alloys, as well as aluminium A99 were used as objects of study. The compositions of the investigated alloys by the results of a spectral analysis are presented in Table 1.[1]

Aluminium 99.99%, master alloys Al–50% Cu and magnesium 99.92% were used as charging materials to prepare alloys. The melt was prepared in an electric resistance furnace in a graphite/fireclay crucible. Ingots 250 mm × 120 mm × 25 mm in size were fabricated by the method of semicontinuous casting, were subjected to homogenization annealing at a temperature of 540 °C (for Al–Cu alloys) and 435 °C (for Al–Mg alloys) to complete dissolution of the nonequilibrium phases of eutectic origin. Ingots were hot rolled to 68% reduction after heating up to 400 °C. Rolled sheets were subjected to softening and cold rolling to 62.5% reduction. The final thickness of cold-rolled specimens was 3 mm. To produce specimens of different grain size, recrystallization annealing was performed at temperatures within the range of 320–540 °C with different durations of isothermal holding: in the case of 30 min and more, in a SNOL-1,6.2,3.0,8/9-M1 furnace; at shorter holding times, in a saltpeter bath. After annealing, the specimens were quenched to provide for the single-phase structure of the alloys.

The mean grain size was measured by a linear intercept method on oxidized microsections. The grain size was varied in the range from about 40 μm to 400 μm for the Al–Mg alloys, while only large-grained Al–Cu specimens with the grain size of about 200 μm and larger were obtained. The concentration of magnesium and copper in solid solution for single-phase alloys was taken to be equal to the weight content of magnesium (copper) in an alloy.

Sheet specimens fabricated according to the ASTM B557M-94 standard were tested for tension at room temperature on a Zwick TC–FR250SN.A4K universal testing machine with automatic recording of tension curves. The tension rate was 4 mm/min which is equivalent to 0.1 s^{-1} strain rate. The curves obtained were used to find values of flow stress at various residual strains within the range of 0.1–20%. The time interval from the end of quenching to the testing of specimens was within half an hour; the mean duration was ∼20 min. In addition, some Al–Cu specimens were tested after two-month natural ageing.

The JEOL 2000EX transmission electron microscope operated at 120 kV was used for the microstructure examination.

3. Experimental results

3.1. Al–Mg alloys

The dependencies of the flow stress, σ, and the work hardening rate, $d\sigma/d\varepsilon$, on the true strain, ε, are given in Fig. 1 for alloys with different magnesium contents (to demonstrate the effect of grain size the data for both large-grain and fine-grain specimens of the same alloys are presented in the figure). It is evident that the concentration of magnesium has a significant effect not only on the level of flow stress, but also on the hardening rate; in the latter case, the character of the effect is more complex. Thus, while the level of flow stress increases approximately linearly with C_{Mg}, the way the

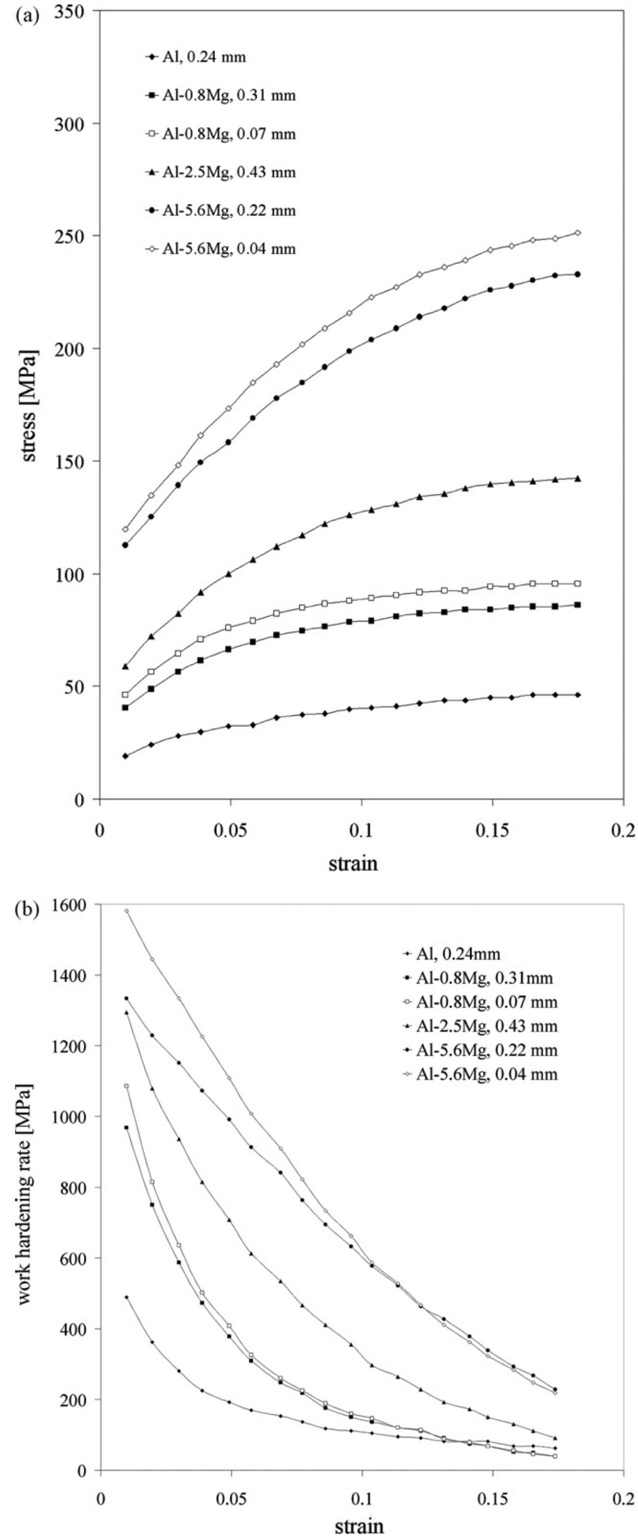

Fig. 1. Effect of magnesium concentration on work hardening of Al–Mg alloys. Dependencies on logarithmic strain are given: (a) for flow stress and (b) for work hardening rate.

[1] In the table and further in this paper the alloying element concentrations are give in atomic percents.