EFFECT OF MAGNESIUM CONTENT ON HOT DEFORMATION AND SUBSEQUENT RECRYSTALLISATION BEHAVIOUR OF ALUMINIUM-MAGNESIUM ALLOYS

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ABSTRACT: Four high purity aluminium-magnesium alloys with different magnesium contents (0.13%, 1%, 3% and 5%Mg) were used to investigate the effect of magnesium content on hot deformation and subsequent recrystallisation behaviour. The alloys were deformed in a plane strain compression testing machine at 385°C and at strain rates of 2.5/s and 25/s. The changing strain-rates from 25/s to 2.5/s and from 0.25/s to 2.5/s were also applied. Polarised light optical microscopy was used for determining recrystallisation kinetics, *i.e.* the time for 50% recrystallisation, and recrystallised grain size. Experimental results show that with increasing magnesium content, flow stress increases, recrystallised grain size decreases continuously, while recrystallisation kinetics increase from the lowest magnesium alloy (Al-0.13%Mg) up to Al-1%Mg and then decrease.

Keywords: *Al-Mg alloys, plane strain compression test, changing strain-rate, flow stress, recrystallisation*

1. INSTRODUCTION

Magnesium is one of the most important solution-hardening elements in aluminium alloys and aluminium alloys with different contents of magnesium are the most important matrix materials of a commercial series of aluminium alloys. The effect of magnesium content on hot deformation and subsequent recrystallisation behaviour is not completely clear because of interaction with some other elements in commercial alloys. The present research is to investigate the effect of magnesium content on hot deformation behaviour under constant and changing strain-rate conditions, and subsequent recrystallisation behaviour using four high purity aluminium-magnesium alloys with different magnesium contents.

2. EXPERIMENTAL

Four high purity aluminium-magnesium alloys, A1-0.13%Mg, A1-1%Mg, A1-3%Mg and A1-5%Mg, supplied by Alcan International Limited, Banbury, were in an as-rolled and annealed condition with dimensions of 12 mm in thickness. The compositions of the experimental materials and grain sizes are given in Table 1.

Specimens were machined with dimensions 60x40x10 mm with the "rolling" direction, the same as that in the slabs provided. Each alloy was deformed at a temperature of 385°C and at constant equivalent strain-rates of 25/s and 2.5/s and changing strain-rates from 25/s to 2.5/s or

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Table 1 Composition and grain size (d_0) of the experimental alloys

	Si (%)	Fe (%)	Cu (%)	Mn (%)	Mg (%)	Zn (%)	Ti (%)	B (%)	d_0 (μm)
Al-0.13%Mg (J180F)	<0.01	<0.01	0.001	<0.001	0.13	<0.001	<0.001	<0.001	250±10
Al-1% Mg (J181F)	<0.01	<0.01	0.001	<0.001	0.91	<0.001	<0.001	<0.001	250±10
Al-3%Mg (J182F)	0.005	<0.01	0.001	<0.001	2.99	<0.001	<0.001	<0.001	140±10
Al-5%Mg (J183F)	0.01	<0.01	0.001	<0.001	4.98	<0.001	<0.001	<0.001	84±4

from 0.25/s to 2.5/s to strain of 1. These strain-rate paths are illustrated schematically in figure 1. A graphite lubricant (Acheson 1403) was applied, which gives a friction coefficient of 0.075. Two thermocouples, one of which was placed into the centre of the deformation zone and another into the undeformed zone, were used to measure the temperature. The thermal history was recorded on a datalogger.

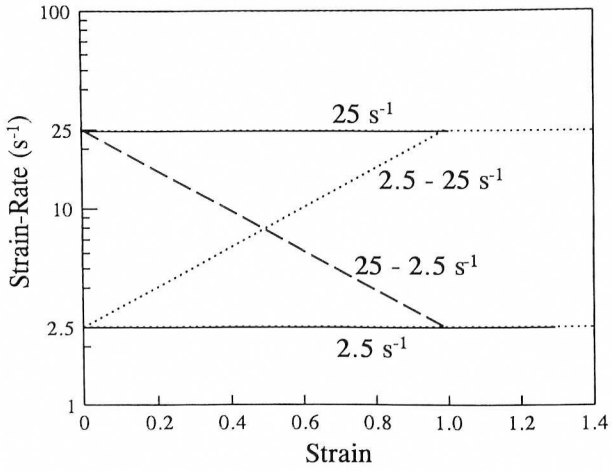

Figure 1 Schematic showing the strain-rate histories used to deform the alloys in the present research.

After deformation, the specimens were immediately drawn out from the PSC furnace and quenched by water spray. The quenched specimens were sectioned and annealed in a salt-bath for different periods. The annealed specimens were prepared by mechanical polishing to 6 μm diamond and then electro-polishing for 60 seconds in a solution of 25% nitric acid + 75% methanol at -25°C and 10-15 volts. The polished specimens were then etched using Barkers reagent (20V for 90sec) for optical examination.

Optical polarised light microscopy was used to determine the recrystallisation kinetics and the recrystallised grain size in the as-polished and etched specimens. The fraction recrystallised and recrystallised grain size were measured by the point counting and the linear intercept method, respectively. The time for 50% recrystallisation (t_{50}), which is used to describe the recrystallisation kinetics, was determined by interpolation of the recrystallisation kinetics curves (fraction of recrystallisation versus annealing time) obtained.

3. RESULTS AND DISCUSSION

3.1 Original grain structure

The as-received materials were fully recrystallised and the average grain size decreased with increasing magnesium content, as shown in table 1. Annealing twins were present in each of the alloys, decreasing in frequency with increasing magnesium content. This may arise from the much higher stored energy in higher magnesium content alloys, which is produced by smaller grain size and larger misorientation across subgrains due to hot working. The above results also indicate that magnesium atoms do not reduce stacking fault energy enough to the formation of annealing twins.

3.2 Flow stress

Figure 2 shows the flow stress-strain curves for the four alloys and figure 3 gives an example of the temperature-strain response for the Al-5%Mg alloy. As expected, the maximum stress increases for a given strain-rate path with increasing magnesium content. As the strain above about 0.2 increases, there is a significant decrease in the stress level in the higher magnesium containing alloys. The reason for this drop in stress level is discussed below.

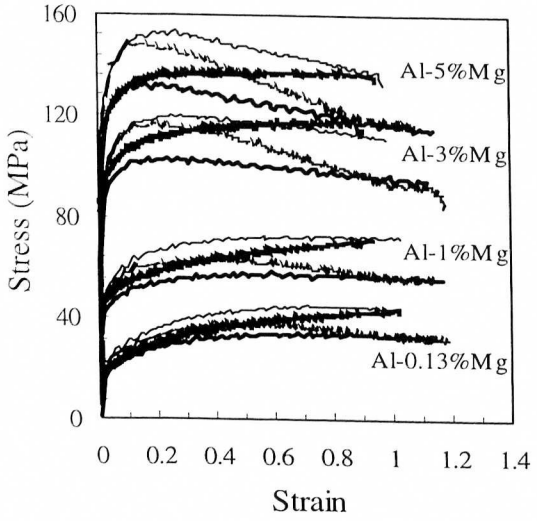

Figure 2 Stress-strain behaviour of the four Al-Mg alloys for various strain-rate histories, showing a mechanical equation of state.

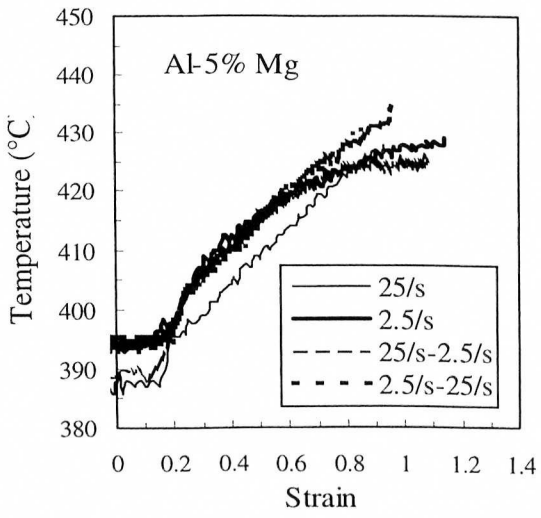

Figure 3 Temperature-strain response during deformation of Al-5%Mg in PSC for various strain-rate histories.

Equation (1) and the experimental data can be used to estimate adiabatic temperature increases for aluminium alloys [1].

$$\Delta T_{max} = \frac{\sigma \varepsilon \beta}{c \rho_D} \qquad (1)$$

where ρ_D = 2690 kg/m^3 is the density of workpiece, c = 933 J/(kg K) is the specific heat of the workpiece and β is the fraction of deformation work converted into heat, typically >0.98 for hot work. Using the mean stress shown in figure 2 in equation (1), the expected temperature increases at a strain of 1 are 50°C and 60°C for the A1-5%Mg specimens deformed at 2.5/s and 25/s,

respectively. These estimated values are higher than the recorded results shown in figure 3, which are 30°C and 40°C, respectively. The lower recorded values can be considered to arise from loss of heat to the cooler tools and to the undeformed regions of the specimens during deformation. The following constitutive equation can be used to estimate the stress drop with the temperature increase,

$$Z = \dot{\varepsilon} \exp\left(\frac{Q_{def}}{RT}\right) = A\left(\sinh(\alpha\sigma)\right)^n \qquad (2)$$

where Q_{def} (= 180 kJ/mol for Al-5%Mg) is the activation energy, R (= 8.31J/mol·K) is the gas constant, T is the temperature (K), α (= 0.052 MPa^{-1}) is a materials constant and n (= 1.59 for Al-5%Mg) is an exponential constant [2]. Using this equation, the stress drop due to temperature increase from 396°C to 426°C is 17 MPa for the deformation condition of 2.5/s. This calculated stress drop is slightly smaller than that measured from the flow curve, which is about 19 MPa, *i.e.* the stress drop in the present research arises mainly from deformational heating, but might not be entirely accounted for. In other words, there might be some microstructural evolution playing a role in the stress drop.

The important feature of figure 2 is that a mechanical equation of state holds for the alloys, *i.e.* the flow stress level is dictated by the instantaneous strain-rate and temperature independent of history. This is the same as found for a commercial Al-1%Mg alloy [3,4]. Figure 4 shows flow stresses at strain of 1 for the four alloys as a function of magnesium content. As expected, the stress increases with increasing magnesium content. For all the alloys, the stresses for the specimens deformed at constant strain-rate of 2.5/s and at decreasing strain-rate from 25/s to 2.5/s are the same, showing the mechanical equation of state. Comparison between the high purity A1-1%Mg and commercial A1-1%Mg shows very close values for all deformation conditions. This indicates that the impurities in A1-1%Mg alloy do not significantly increase the flow stress under the present deformation conditions.

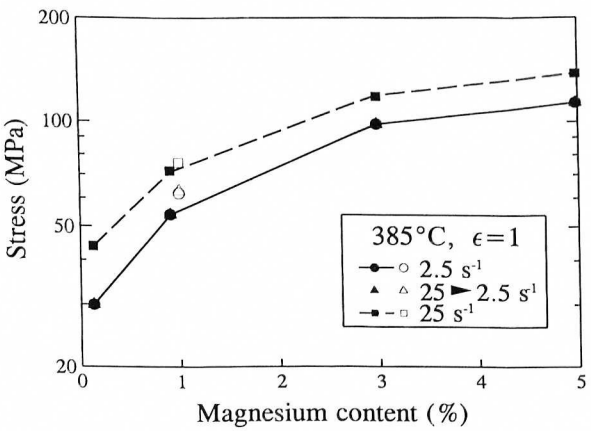

Figure 4 Stress at a strain of 1 as a function of magnesium content for deformation at 385°C. (solid symbols for high purity alloys and open symbols for commercial alloy)

3.3 Recrystallisation behaviour

Figure 5 shows recrystallised grain size (d_{rex}) and the time of 50% recrystallisation (t_{50}) as a function of magnesium content. Recrystallised grain size (d_{rex}) decreases continuously with increasing magnesium content. This arises from the smaller original grain sizes and higher stored energy in higher magnesium content alloys. For nucleation in the specimens deformed to a larger strain than (say) 0.2, it can be assumed that the nucleation is site-saturated and dominated by grain boundary surface, and the recrystallised grain size can then be written as [5]

$$d_{rex} \propto N_V^{-1/3} \propto p^{-1/3} S_V^{-1/3} \delta^{2/3} \quad (3)$$

where N_V is the nucleation density of recrystallisation, p is the probability that subgrains are larger than the critical size for nucleation of recrystallisation and S_V is the grain boundary area per unit volume, which can be calculated from the original grain size (d_0). The probability p is determined by the stored energy, which can be calculated using the values for the internal state variables (ρ_i, δ, θ) and the distribution of subgrain size [5,6]. S_V can be calculated in two cases, *i.e.* equiaxed and elongated grains. For equiaxed grains $S_V \propto 1/d_0$. Using the original grain sizes given in Table 1, the normalised recrystallised grain sizes, which is defined here as the ratios $d_{rex}/d_0^{1/3}$, are 25 in Al-0.13%Mg, 20 in Al-1%Mg, 14 in Al-3%Mg and 11 in Al-5%Mg, respectively, for the specimens deformed at 2.5/s and 385°C to strain of 1, decreasing continuously with increasing magnesium content. $d_{rex}/d_0^{1/3}$ is mainly determined by dislocation substructure. Figure 5a also shows that recrystallised grain sizes in high purity Al-1%Mg are larger than in commercial Al-1%Mg for the different deformation conditions. Again using experimental data given in [3,4,5] for commercial Al-1%Mg, where original grains are slightly elongated with sizes of d_{01} = 115 μm, d_{02} = 75 μm and d_{03} = 94 μm, the grain boundary area per unit volume can be calculated using S_v = $0.429/d_{01} + 2.571/d_{02} - 1/d_{03}$ [7], and the normalised grain size for the specimen deformed at 2.5/s is 15. This value is smaller than that in the high purity Al-1%Mg alloy. This can be considered to arise from the higher stored energy in the commercial purity alloy due to the smaller subgrain size and larger misorientation [6] through the probability term in equation (3) in the commercial alloys.

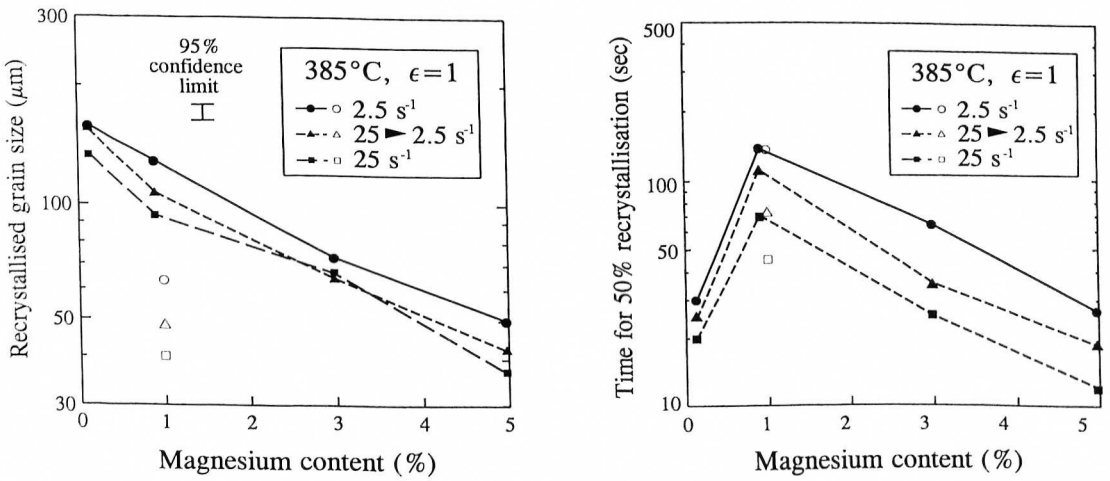

Figure 5 (a) recrystallised grain size and (b) the time for 50% recrystallisation in the specimens deformed at 385°C to strain of 1 and then annealed at 385°C as a function of magnesium content. (solid symbols for high purity and open symbols for commercial alloys)

With increasing magnesium content, t_{50} increases up to 1%Mg and then decreases. No significant difference of t_{50} in high purity and in commercial Al-1%Mg is found for the specimens deformed at 2.5/s, but higher values in high purity than in commercial Al-1%Mg for 25/s. Nucleation density of recrystallisation is controlled by stored energy and distribution of subgrain size, which in turn determines recrystallisation kinetics in combination with the growth rate of the nuclei. The relationship between t_{50}, N_V (or d_{rex}) and growth rate of the nuclei at 50% recrystallisation (G_{50}) can be written as,

$$t_{50} \propto \frac{N_V^{-1/3}}{G_{50}} \propto \frac{d_{rex}}{G_{50}} \qquad (4)$$

Using t_{50} and d_{rex} shown in figure 5, G_{50} is larger in A1-0.13%Mg than other higher magnesium content alloys by a factor of about 5.6, 4.6 and 2.8 for Al-1%Mg, Al-3%Mg and Al-5%Mg, respectively, deformed at 385°C and 2.5/s to strain of 1. The decrease in growth rate of nuclei with magnesium content up to 1%Mg can be considered to arise from the lower mobility of grain boundaries due to viscous drag by magnesium, while the increase with magnesium content above 1%Mg may arise from the much higher stored energy, which compensates the effect of viscous drag by magnesium atoms on mobility of grain boundaries. Comparison of growth rate in high purity and commercial alloys can also be conducted using the results obtained in [3,4,5,6]. The growth rate in high purity Al-1%Mg is larger than in commercial Al-1%Mg by a factor of about 2. The combination of higher growth rate of the nuclei and lower stored energy in high purity Al-1%Mg alloy, which arises from the larger subgrain size and smaller misorientation, gives similar values of t_{50} in both high purity and commercial Al-1%Mg alloys.

4. CONCLUSIONS

- Flow stress increases with increasing magnesium content for constant strain rate and decreasing strain rate deformation. A mechanical equation of state holds for all the Al-Mg alloys.
- No significant difference of flow stress between commercial and high purity Al-1%Mg alloys is observed under the applied deformation conditions.
- The time for 50% recrystallisation increases initially up to 1%Mg and then decreases with increasing magnesium content, while recrystallised grain size decreases continuously with increasing magnesium content.

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