PARTICLE COARSENING AND SUBGRAIN GROWTH IN A TWIN ROLL CAST AA8006 ALLOY

Børge Forbord*, Ole Daaland and Erik Nes***

* Dept. of Metallurgy, The Norwegian University of Sci. and Tech., N-7034 Trondheim, Norway
**Hydro Aluminium, R&D Materials Technology, N-6285 Håvik, Norway

ABSTRACT A low-eutectic twin roll cast and cold rolled AA8006 aluminium alloy (Al1.8%Fe0.8%Mn0.2%Si) has been examined in TEM in order to follow the softening process during back-annealing. The alloy is characterised by a high volume fraction of finely dispersed intermetallics, mainly situated on the subgrain boundaries. Cold rolling to a strain $\varepsilon=2$ results in a nearly equiaxed, highly developed subgrain structure with a mean sub-boundary misorientation of about 5°. The behaviour during annealing has been followed with respect to the variations in subgrain size and particle size. The sub-boundary structure is stabilised by the intermetallic particles. As a consequence recrystallisation is delayed, leading to an almost complete softening only due to recovery. Studies of particle coarsening in recrystallised material have shown that volume diffusion contributes very little to the coarsening. It can therefore be concluded that in these type of alloys particle coarsening is mainly controlled by other diffusion mechanisms such as HAGB- or LAGB-diffusion.

Keywords : *particle coarsening, subgrain growth, coupled growth, diffusion mechanisms, growth kinetics*

1. INTRODUCTION

A twin roll cast AA8006 aluminium alloy has been investigated. Twin roll casting involves large solidification rates and results in a fine distribution of second phase particles but also a supersaturation of slowly diffusing elements such as Fe and Mn. The alloy as rolled contained a second phase volume fraction of approximately 8%. In sheet metal production the final properties are controlled by annealing after cold rolling (back-annealing). During the anneal of the AA8006-alloy the intermetallics play a significant role in controlling recovery and recrystallisation. A Zener-drag will act upon the moving subgrain boundaries, and prevent the subgrains from becoming potential nuclei for recrystallisation. As a consequence the fine-grained structure formed during cold rolling is practically maintained during back-annealing. This results in a material with an excellent combination of strength and ductility.

Morris [1] studied an Al6%Ni-alloy which in microstructural terms was similar to the AA8006 alloy investigated here. The subgrain coarsening in Morris' alloy is shown in figure 1, and it can be seen that the subgrain size changes linearly with the cube root of time. This has also been found previously for both the subgrain growth and the particle coarsening in AA8006 [2], and new results also confirm these relationships (Figure 2). In both cases it was concluded that the subgrain growth was governed by the volume diffusion controlled coarsening of the particles. This conclusion is not entirely correct, and the main objective of this work has been to find a more correct interpretation of the results. The very complex nature of these type of alloys makes this very difficult.

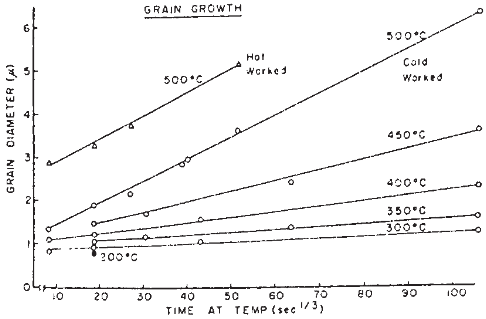


Figure 1 : Subgrain growth in Morris' Al6wt%Ni alloy.

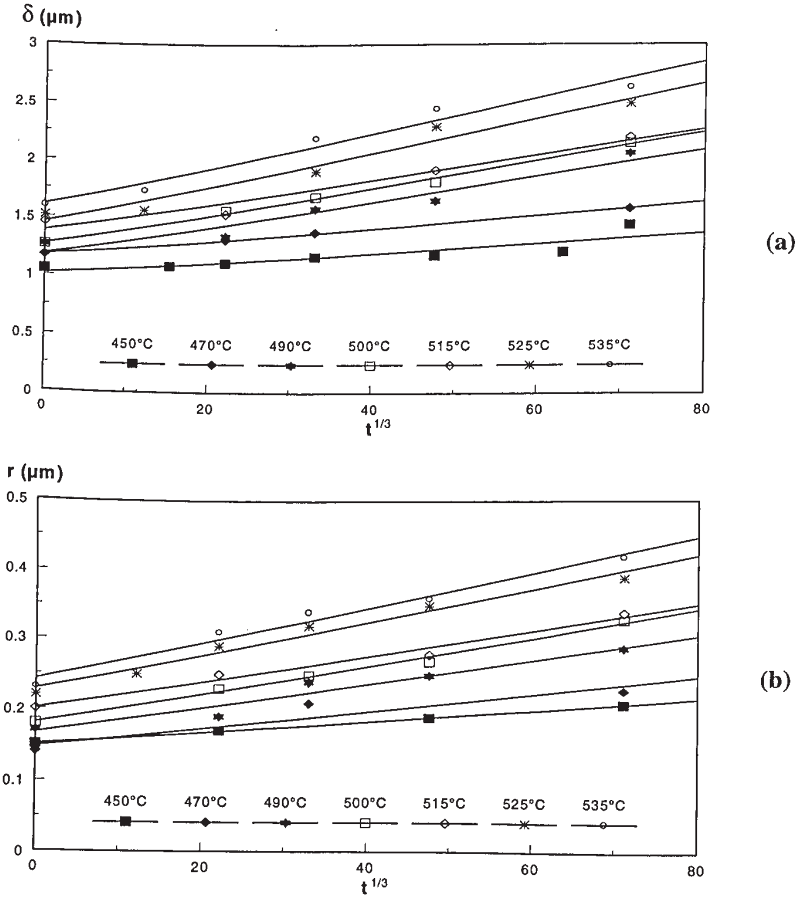


Figure 2 : a) Subgrain growth and b) particle coarsening in AA8006.

2. EXPERIMENTAL

The composition of the AA8006 alloy was Al1.8wt%Fe0.8wt%Mn0.2wt%Si, and the strip thickness of the material as cast was 4 mm. The alloy was first cold rolled to a thickness of 2 mm. It was then batch-annealed to 475°C, applying a heating rate of 50°C/h, and held at that temperature for 3 hours. During this intermediate anneal the supersaturated solid solution precipitated as finely distributed intermetallics on the substructure; i.e. the second phase volume fraction increased considerably. Finally the alloys were cold rolled from 2 mm to approximately 0.3 mm (ε=2.7).

The softening process was studied by preparing thin foils for TEM from specimens heated at a constant rate (50°C/h) to seven different temperatures, and then held at these temperatures for different time periods. The thin foils were prepared by subjecting the specimens to jet electropolishing on a Struers Tenupol, using an electrolyte containing two parts of methanol and one part of nitric acid. Electropolishing was performed at 20V with an electrolyte temperature between -30°C and -20°C. Subgrain and particle growth kinetics were studied on a JEOL JEM-2010. The subgrains were measured by linear interception in the rolling direction from TEM-micrographs. Particle radii was measured from micrographs using a Kontron Videoplan. At least 150 subgrains and 400 particles have been measured in each specimen.

3. RESULTS AND DISCUSSION

In alloys with a high volume fraction of second phase particles a situation can arise where grain growth has stagnated due to the particle dispersion. Subgrain growth will now proceed as a result of the coarsening of the intermetallics. The rate of growth is controlled by the rate of change of particle size :

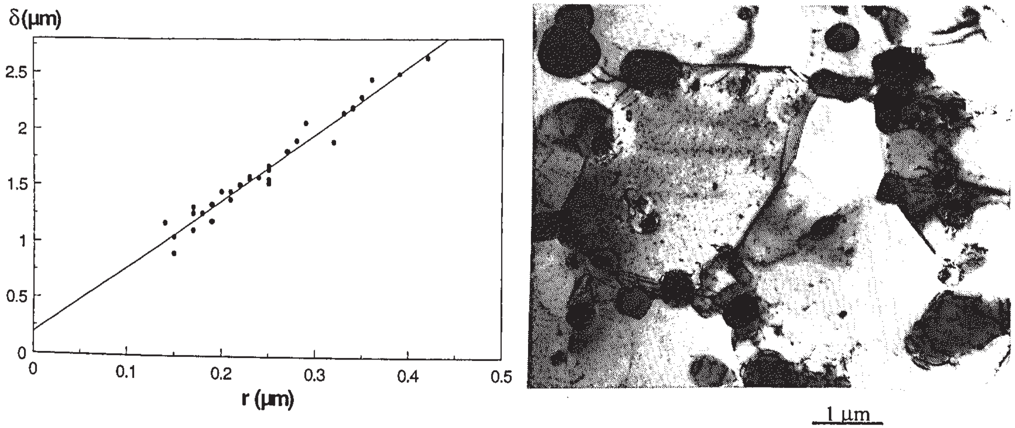
$$\frac{d\delta}{dt} = c \cdot \frac{dr}{dt} \tag{1}$$

where δ is the subgrain size, r is the particle size and c is a proportionality constant. Figure 2 shows that the subgrains and particles in AA8006 grow in the same manner. This indicates that the subgrain growth in the alloy is governed by the coarsening of particles, and figure 3a confirms that this is the case. Proportionality between the subgrain size and particle size has also been found in other experiments. Humphreys and Chan [3] found a linear relationship between these two parameters in an Al6wt%Ni alloy containing a volume fraction of 0.10 $NiAl_3$ particles. Yang et. al [4] found the same for silver particles in a Nickel matrix.

The rate of particle coarsening will again depend on the rate controlling mechanism [5, 6, 7, 8]. A general expression for the coarsening is given in the Lifshitz, Slyozov and Wagner (LSW)-theory [9, 10] :

$$r^n - r_0^n = K \cdot t \tag{2}$$

where r is the particle size at time t, and K is the growth rate constant. In case of volume diffusion of solute to the particle n=3, while if boundary diffusion is rate controlling n usually equals 4. Under certain circumstances for low angle grain boundaries n has even been found to be 5 (pipediffusion). In the AA8006 alloy studies of coarsening in recrystallised material has shown that the increase in particle size is small (Table 1). Coarsening is negligible when the material is annealed at 450°C and 500°C. At 525°C there is some increase in particle size though, indicating that there is a small contribution from volume diffusion at this temperature. As a result it can be concluded that volume diffusion is not a dominant rate controlling mechanism. This means that the exponent in equation (2) in this case should be 4 or 5. The fact that most of the particles are situated on the subgrain boundaries (Figure 3b) also support an n-value of 4 or 5.



(a) (b)

Figure 3 : a) The linear relationship between the subgrain size and particle size, b) TEM-micrograph showing particles on the subgrain boundaries and nodes in the AA8006-alloy.

Table 1 : Investigation of particle growth in recrystallised AA8006.

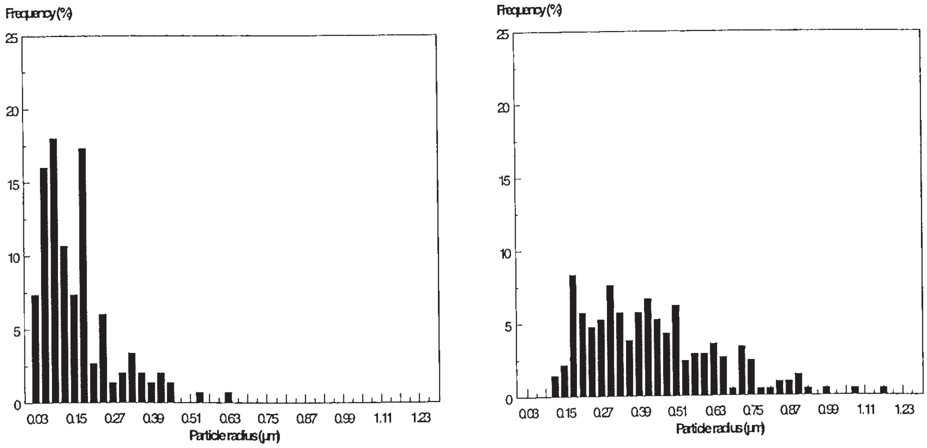
Annealing time [s]	Particle radius (µm)		
	450°C	500°C	525°C
10	0.11	0.14	0.15
180			0.17
36 000	0.11	0.16	0.20
108 000			0.23
360 000	0.14	0.17	0.26

However, the n-value has proved very difficult to find in this alloy. The LSW-theory also requires a stationary particle size distribution (PSD) in order to apply. Figure 4 clearly shows that this is not the case here, as the PSD changes its shape during annealing. As a consequence n-values of 4 or 5 also fails to describe the coarsening in AA8006. Figure 5 shows that in both cases there is a significant deviation from linearity if the data are plotted as δ^4-δ_0^4 vs. t (Fig. 5a) or δ^5-δ_0^5 vs. t (Fig. 5b), as the coarsening proceeds faster during the early stages of annealing. A similar interpretation of Morris' results in Al6wt%Ni does not give a satisfactory explanation either.

This leaves us with the linear relationship between δ and $t^{1/3}$

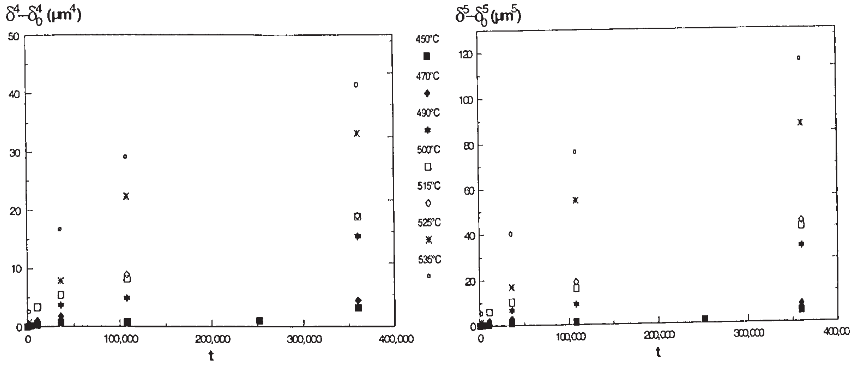
$$\delta = \delta_0 + K \cdot t^{1/3} \tag{3}$$

It must be emphasized, though, that this relationship is purely empirical. According to Morris this relationship indicated that the subgrain growth were governed by the volume diffusion controlled coarsening of particles. However, equation (2) shows that according to the coarsening theory, this way of plotting the data is only correct when $\delta >> \delta_0$. In this case where the coarsening is limited, we cannot neglect δ_0. Another problem with this way of plotting the data is that it is dependent on which time is rated as "zero time" in figures 1 or 2. If zero time was changed, the slopes of the curves would change. This will in turn affect the calculation of activation energies.



(a) (b)

Figure 4 : PSD in AA8006 back-annealed to a) 450°C and b) 535°C+100 hours at temperature.



(a) (b)

Figure 5 : LSW-plots with a) n=4 and b) n=5.

It is therefore rather puzzling that a fairly reasonable activation energy is obtained when using the empirical relationship in equation (3). For the AA8006 alloy it was found to be ~220 kJ/mol and for Morris' Al6%Ni alloy it was ~160 kJ/mol [2]. These values corresponds to the activation energies for diffusion of Mn and Ni in aluminium respectively. However, table 1 showed that coarsening controlled by diffusion through the volume was small compared to the actual coarsening. In the case of boundary diffusion when the diffusion path is more open, an activation energy of about half these values would be expected.

The very complex nature of these alloys makes it very difficult to model the coarsening in a proper manner. It is very likely that the different coarsening mechanisms operate simultaneously, with boundary diffusion as the dominant coarsening mechanism.

4. CONCLUSIONS

1) The subgrain growth in the alloy is governed by the coarsening of particles.
2) Diffusion of solute to the particles occurs mainly by boundary diffusion.
3) LSW-theory does not apply in either Morris' Al6wt%Ni or AA8006. This may be due to the non-stationary particle size distributions in these alloys. To develop a proper model for the coarsening has proved to be very difficult.
4) The empirical relationship in equation (3) can be used to describe the coarsening in both alloys, but its weaknesses should be noted.

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