STRENGTHENING MECHANISMS IN A TWIN ROLL CAST AA8006 ALLOY

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ABSTRACT A twin roll cast AA8006 aluminium alloy (Al1.8%Fe0.8%Mn0.2%Si) containing a high volume fraction of second phase particles has been investigated. Cold rolling to a strain ε=2 results in a nearly equiaxed fine-grained structure, with the particles mainly situated on the subgrain-boundaries. Zener-drag causes recrystallisation to be delayed, and the alloy softens almost completely due to recovery. This gives a material with an excellent combination of strength and ductility. The coarsening of the particles and the growth of subgrains have been followed in detail in TEM. Tension tests have shown that the flow stress varies linearly with the inverse subgrain size ($\alpha_l \simeq 2.9$). One of the objectives of this work has been to seperate the strengthening mechanisms in the alloy, i.e. seperate the contributions due to particles and subgrain boundaries. By subjecting specimens back-annealed to different levels of strength (different subgrain sizes) to a flash-annealing treatment at high temperatures, they will recrystallise. In the recrystallised material there will only be a contribution from the second phase particles and solid solution.

Keywords : *strengthening mechanisms, subgrain strengthening, particle hardening*

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. In sheet metal production the final properties are controlled by annealing after cold rolling (back-annealing). Cold rolling results in a fine-grained structure with an average grain size of approximately 0.45 µm. During annealing of the AA8006-alloy intermetallics ($f \approx 0.08$) 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] investigated an Al6wt%Ni-alloy which in microstructural terms was similar to the present alloy. The effect of grain size on the flow stress was examined. He found a linear relationship between these two parameters with α_l=6.7 (see equation (1)) which is surprisingly high. This is shown in figure 1. Sæter [2] found α_l=2 and this value seems to be more reasonable. The contribution from the particles to the flow stress in Morris' work was estimated to be approximately 20 Mpa, which is a relatively small contribution. The objective of this work has been to separate the contribution to the flow stress from particles and subgrains in AA8006, and to establish a relationship between the flow stress and microstructural parameters.

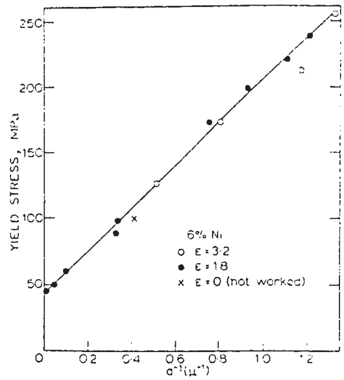

Figure 1 : The relationship between the flow stress and the inverse subgrain size in Morris' Al6wt%Ni-alloy.

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 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 alloy were cold rolled from 2 mm to approximately 0.3 mm (ε=2.7). Back-annealing of specimens by continuous heating to different levels of strength gave a relationship between the strength and the inverse subgrain size. Some of the specimens were subjected to a subsequent flash-annealing treatment in order to isolate the contribution from the particles and solid solution. This has made it possible to study the relationships between the strength and the microstructural parameters.

The flow stress measurements were found from tension tests, in which the 0.2% offset was defined as the yield point. 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 sizes 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. Misorientation measurements were carried out using a Phillips EM 400 with equipment for on line analysis of grain orientations. The analysis is performed using a convergent beam technique which allows the Kikuchi pattern of every distinct grain to be imaged. A simple video recording system is used to digitalise the image and a convenient computer software is used to index the Kikuchi patterns.

3. RESULTS AND DISCUSSION

Low angle boundaries, high angle boundaries and particles increase the resistance to dislocation motion and hence increase the flow stress of most materials. This strengthening has been described by empirical relationships such as

$$\sigma = \sigma_i + \sigma_p + \alpha_1 \cdot M \cdot G \cdot b \cdot \delta^{-n} \qquad (1)$$

where σ is the flow stress, σ_i is the frictional stress due to elements in solid solution, σ_p is the particle contribution, α_1 is a constant, the Taylor factor M=3.1, the shear module G=26.5 GPa, the Burger's vector $b=2.86 \cdot 10^{-10}$ m. δ is here the subgrain size determined from linear interception in the rolling direction. When n=1/2 equation (1) becomes the well-known Hall-Petch formula [3]. For fine-grained materials containing low-angle boundaries several authors have found n=1 [2, 4, 5, 6, 7]. As already shown this was also found by Morris in Al6wt%Ni.

In the present experiments the alloy has been annealed for various times at different temperatures. These treatments have provided a range of subgrain sizes as well as particle sizes. Subgrain growth is controlled by particle coarsening as explained in another paper [8]. The linear relationship between subgrain size and particle radius is given below in figure 2. It follows that the particle contribution cannot be treated as a constant in equation (1).

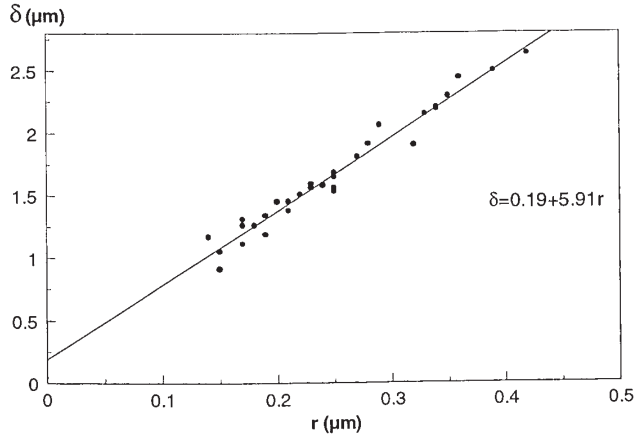

Figure 2 : The linear relationship between the subgrain size and particle radius showing that subgrain growth is controlled by the coarsening of particles.

However, table 1 shows that the particle contribution in the early stages of back-annealing is relatively constant. In this alloy a stabilisation treatment is necessary in order to avoid recrystallisation at high temperatures. The structure is stabilised by applying a low heating rate, in which some precipitation plays an important role. During these early stages particle coarsening is small; i.e the particle contribution does not change much. A value for σ_p is now found by subjecting specimens to a flash-annealing treatment. By doing this the specimens recrystallise and the contribution to the strength from the particles and solid solution are isolated. It can be seen in the table that the value of $\sigma_i+\sigma_p$ is approximately 65 MPa in all the investigated specimens. If we assume that $\sigma_i \simeq 30$ MPa this leaves a contribution of about 35 MPa from the particles. From this value an effective volume fraction of particles can be estimated based on the following expression for the Orowan-stress :

$$\sigma_p = \frac{M \cdot G \cdot b}{\lambda} \tag{2}$$

where the particle separation in the glide plane is given by $\lambda=0.8[(\pi/f)^{1/2}-2)r$. From these two expressions an effective volume fraction of 5% is obtained, which is a reasonable value compared to the measured value ($f \approx 0.08$). The particles display a log-normal size distribution, which means that the effective f becomes somewhat lower. This value has been used in the following calculations where we have a situation where the subgrains and particles grow in a coupled manner; i.e. the contribution from the particles changes continuously.

Table 1 : Flow stress before and after flash-annealing in specimens back-annealed to a certain temperature.

Temperature	$R_{p,0.2}$ - before flash-annealing	$R_{p,0.2}$ after flash-annealing
As rolled to ε=2	204.0 MPa	66.8 MPa
225°C	179.5 MPa	65.0 MPa
325°C	170.0 MPa	65.3 MPa
375°C	155.4 MPa	64.3 MPa
450°C	127.0 MPa	62.7 MPa
500°C	106.0 MPa	61.2 MPa

On the effect of the substructure on flow stress, let us first investigate the Hall-Petch relationship. Such a relationship can be ruled out immediately, see figure 3a. Although there is a linear relationship between σ and $1/\delta^{1/2}$, the best-fit-straight-line intersects the y-axis at a negative value. This is physically unrealistic as our measurements have shown that $\sigma_i+\sigma_p$ should be between 30 and 70 MPa. As a consequence such a relationship can be ruled out.

In another approach (one-parameter analysis) it is often assumed that the flow stess scales with the total dislocation density with a relationship which can be formulated as follows :

$$\sigma = \sigma_i + \alpha_2 \cdot M \cdot G \cdot b \cdot \sqrt{\rho_{tot}} \tag{3}$$

where α_2 is a constant and ρ_{tot} is the total dislocation density. The flow stress is here expected to be controlled by the dislocation density alone. This means that their distribution is not assumed to have any influence on the strength. For deformed aluminium, the dislocations are observed to be organised in a cell/subgrain structure defined by its size δ and the misorientation θ. Consequently the total dislocation density can be estimated by

$$\rho_{tot} = \rho_i + \frac{\kappa \cdot \theta}{b \cdot \delta} \tag{4}$$

where ρ_i is the dislocation density in the cell/subgrain interiors and κ is a geometrical constant in the order of 3. If we assume that the interior dislocations only make a small contribution to the total density, equation (4) reduces to $\rho_{tot} = (\kappa\cdot\theta)/(b\cdot\delta)$. By inserting this reduced expression into equation (3) the flow stress will now be given as

$$\sigma = \sigma_i + \alpha_2 \cdot M \cdot G \cdot b \sqrt{\frac{\kappa \cdot \theta}{b \cdot \delta}} \tag{5}$$

The flow stress as a function of $(\theta/\delta)^{1/2}$ is shown in figure 3b, and the values in the plot are given in table 2. It should be noted that θ in this plot is in radians. Even though there are just a few points in this figure, it clearly indicates that this way of interpreting the data is not correct. In this plot α_2=0.3 which is reasonable [9], but the best-fit-straight-line once again intersects the y-axis at a negative value.

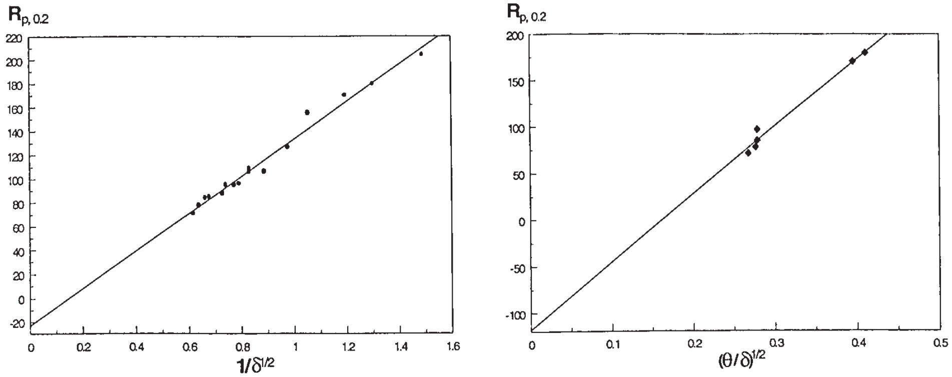

Figure 3 : a) The Hall-Petch relationship and b) the flow stress as a function of $(\theta/\delta)^{1/2}$ for various deformation and annealing conditions.

Table 2 : Flow stress, misorientations and subgrain sizes in AA8006.

Flow stress (MPa)	Misorientation (°)	Subgrain size (μm)
179.5	5.7	0.59
170.0	6.3	0.70
96.3	7.1	1.59
84.7	9.8	2.19
77.7	10.7	2.44
71.0	10.8	2.63

An n-value of 1 seems to be more realistic. The flow stress vs. the inverse subgrain size is plotted in figure 4, and it can be seen that there is a discontinuity in the plot. This is due to the fact that for $1/\delta>0.8$ σ_p is relatively constant, while for $1/\delta<0.8$ it is not. Let us first consider the first case when $\sigma_i+\sigma_p=65$ MPa. The best-fit-straight-line drawn from this value through the points gives an α_1-value of 2.9. Inserting these values into equation (1), the mathematical relationship between the yield stress and the microstructural parameters now becomes :

$$\sigma[MPa]=\sigma_i+\sigma_p+\alpha_1\cdot M\cdot G\cdot b\cdot\frac{1}{\delta}=65+2.9\cdot M\cdot G\cdot b\cdot\frac{1}{\delta} \tag{6}$$

During the later stages of annealing ($1/\delta<0.8$) the situation changes. The particles and subgrains grow in a coupled manner where $\delta=0.19+5.91r$ and large particles grow at the expense of the smaller ones. Rearranging and inserting this relationship in the expression for λ, gives us the following contribution from the particles in the alloy :

$$\sigma_p=\frac{M\cdot G\cdot b}{0.8\cdot\left[(\pi/f)^{1/2}-2\right]\left[0.17\cdot\delta-0.03\right]} \tag{7}$$

This equation can be reduced to $\sigma_p\approx1.3\cdot M\cdot G\cdot b/\delta$, and the mathematical relationship between the flow stress and the microstructural parameters for $1/\delta<0.8$ then becomes

$$\sigma[MPa]=\sigma_i+\sigma_p+2.9\cdot M\cdot G\cdot b\cdot\frac{1}{\delta}=30+4.2\cdot M\cdot G\cdot b\cdot\frac{1}{\delta} \tag{8}$$

The slope of this curve (4.2 MGb) is well below the one found by Morris. There are two probable reasons for this. In Morris' work $\delta=\sqrt{A}$. This will increase the α_l-value by a factor of 1.2-1.5 compared to measuring by linear intercept. Secondly, the low estimation of the contribution from particles (20 MPa) was taken from the flow stress in extensively annealed material. However, the extensive coarsening of intermetallics in Morris' alloy will have a significant effect on the flow stress (see equations (7) and (8)). Morris' value is therefore not unreasonable after all.

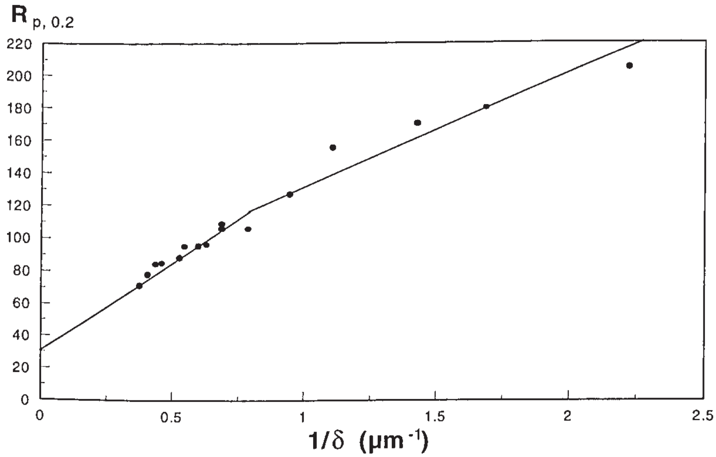

Figure 4 : The linear relationship between the yield stress and the inverse subgrain size in the AA8006 alloy.

4. CONCLUSIONS

1. The flow stress changes linearly with the inverse subgrain size (α_l=2.9)
2. For $1/\delta>0.8$ (during stabilisation) the particle contribution is reasonably constant. For $1/\delta<0.8$ the particle contribution changes as large particles grow at the expense of smaller ones.
3. The Hall-Petch relationship (n=1/2) and a one-parameter analysis have also been investigated, but neither of these two theories could satisfactorily explain the softening in the alloy.

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ACKNOWLEDGEMENT

Thanks are due to Hydro Aluminium, Karmøy Rolling Mill a.s for financial support, permission to publish this paper and for supply of the alloy.