EFFECT OF ORIGINAL GRAIN SIZE ON STATIC RECRYSTALLISATION OF Al-1Mg DURING HOT ROLLING

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Abstract It is well known that grain boundary regions provide potent nucleation sites for static recrystallisation. This fact influences both the resultant grain size and texture developed, which are essential to product quality control. However, the limited studies on the effect of grain boundaries on the kinetics of static recrystallisation for aluminium alloys have shown a spuriously high effect of original grain size. Multipass hot rolling tests of initially tapered slabs of an Al-1Mg alloy were carried out in order to investigate the effect of original grain size on the kinetics of static recrystallisation by using a quantitative metallographic technique. The experimental results have been compared with prior work on a similar Al-1Mg alloy tested in plane strain compression. The results also suggest that grain boundary regions are important nucleation sites at low strains, but high strains activate intragranular sites.

Keywords: *grain boundaries, nucleation sites, recrystallisation, thermomechanical processing.*

Introduction

The effect of original grain size on the kinetics of static recrystallisation has not been resolved adequately for aluminium alloys, probably because of the difficulty in obtaining structures of different grain size without producing significant changes in texture and/or alloy constitution. Basic principles indicate that for grain boundary nucleated transformations [1]

$$t_{0.5} = \alpha \cdot d_0^a \tag{1}$$

$$d_{rex} = \beta \cdot d_0^b \tag{2}$$

where $t_{0.5}$ is the time for 50% recrystallisation, d_{rex} the recrystallised grain size and d_0 the original grain size. The exponents a and b depend on the type of preferential nucleation site. Previous research [2] had determined the kinetics of recrystallisation of an Al-1Mg alloy after deformation 2, when complete recrystallisation occurs after deformation 1 in plane strain compression (PSC) at 400°C. Different grain sizes were obtained by varying the prior thermomechanical history. However, when the relationships obtained were employed for modelling hot rolling, unrealistically coarse grain structures were predicted, and this was shown to be due to the original grain size effect [3]. This paper deals with the effect of original grain size on static recrystallisation determined by using double deformation tests on originally tapered slabs, aiming at minimising alloy constitution and/or texture changes.

Experimental Procedure

The chemical composition of the Al-1Mg alloy employed in the present work is shown in Table 1. This alloy was prepared as part of a major Brite-Euram project [4], whereas the geometry of the tapered slabs has been shown elsewhere [5].

Table 1 . Chemical composition (wt%) of the Al-1Mg alloy used in the present work.

Al	Mg	Mn	Fe	Si	Others*
98.56	0.94	0.001	0.31	0.15	0.039

* It includes small quantities of Cu, Ca, Na, Ti and B.

In order to determine the effect of original grain size on static recrystallisation, tapered slabs were used, and processed according to tests I and II. Table 2 shows a detailed description of such double deformation rolling tests, which were carried out in a fully instrumented two-high 50 tonne Hille mill, with two electrically-heated muffle type furnaces situated to either side to perform annealing treatments.

Table 2. Experimental matrix for double deformation tests on tapered slabs indicating the processing parameters.

Test	Reheat T (°C)	Roll T_1 (°C)	ε_1	Anneal T_1 (°C)	Anneal t_1 (s)	Roll T_2 (°C)	ε_2	Q	Anneal T_2 (°C)	Anneal t_2(s)
I	510	505	≤0.71	450	35	450	0.3	No	400	Various, Q
II	510	505	≤0.71	450	35	450	0.71	No	400	Various, Q

Q: Cold water quench.

Tapered slabs were employed because when they are rolled, their sections (analysis planes) receive different strains, leading on annealing to different recrystallised grain size along their length. The analysis plane in the tapered slab leading to the coarser grain size (113 ± 5 μm) received a strain of 0.4, while the strain producing the finer grain size (58 ± 3 μm) was 0.71. A second rolling pass and subsequent annealing thus enables the effect of original grain size to be studied in selected planes. In the present case, two strains were applied upon the second pass, namely, 0.3 and 0.71. Samples were prepared metallographycally for fraction recrystallised and recrystallised grain size (mean linear intercept) measurements. The samples were ground on successively finer silica papers and mechanically polished with diamond compounds. The final polishing was carried out with a colloidal suspension of SiO_2 in water. After polishing, the samples were anodised in Barker's etch at 20 V DC. An optical microscope equipped with a micrometer stage and a point counter device was used to perform the measurements while the sample was observed under crossed polarisers.

However, for the recrystallisation behaviour to be associated solely to original grain size changes, no changes in alloy constitution and/or texture should occur. Hence, an image analyser attached to an optical microscope was employed to observe changes in alloy constitution, whereas X-ray diffraction was used to measure the recrystallisation texture. Both type of measurements were carried out in two sections, the one leading to the coarser grain size and that leading to the finer grain size. Samples for particle size analysis were ground, polished and etched with HF before measurements of volume fraction and particle size. The samples for bulk texture measurements, which were carried out at Alcan International Ltd, Banbury Laboratory, UK, were 25x25x10 mm in size, cut in such a way as to make the irradiated surface parallel to the slab surface in contact with the rolls. They were later ground in successively finer silica papers, polished down to 1 μm finish

using industrial diamond compounds and etched with HF to remove the deformed surface layer, prior to irradiation.

Results and Discussion

Table 3 shows the recrystallisation texture measured at the coarse and fine grained regions. These components are typical of the recrystallisation texture in aluminium alloys [6]. The random component increases some 25% from the coarse to the fine region, which was expected as a fine grain size tends to randomise the texture. There are also important changes in the Brass and S components, which may be related to competition of different nucleation sites. However, notice that both cube and rotated cube components do not change much. Hence, these texture changes are not expected to effect significantly static recrystallisation.

Table 3. Volume fraction of texture components for coarse and fine grain size regions of originally tapered slabs after rolling pass 1 (Tests I and II, Table 2), measured by X-ray diffraction.

Texture Component	Volume fraction (%)	
	Coarse region (113 μm)	Fine region (58 μm)
Cube {001}<100>	8.74	7.96
Brass {011}<211>	10.63	6.20
Copper {225}<554>	5.86	5.53
Goss {011}<100>	5.57	3.02
S {123}<634>	15.70	12.30
H (45° ND rotated cube)	4.49	6.29
CH (17° ND rotated cube)	13.45	14.50
Random	35.56	44.20

As far as alloy constitutional changes are concerned, Table 4 presents volume fraction and particle size of Mg_2Si/Others and $AlFe_3$ for both coarse ($\varepsilon = 0.4$) and fine ($\varepsilon = 0.71$) grain size regions, also including for comparison the figures in the as-received material and the region that received a strain $\varepsilon = 0.2$, which was partially recrystallised.

Table 4. Particle size analysis. Initial deformation temperature 505°C; annealing temperature 450°C, for 35 s. Size defined as a mean sphere diameter, p (μm); volume fraction f_v in %. The measurements were carried out half way between sample surface and centre.

Strain at slab centre	Mg_2Si (plus others)		$AlFe_3$	
	f_v (%)	p (μm)	f_v (%)	p (μm)
As-received	0.13 ± 0.01	1.38 ± 0.10	1.73 ± 0.02	1.70 ± 0.07
0.20	0.07 ± 0.01	1.13 ± 0.07	1.84 ± 0.05	1.83 ± 0.13
0.40	0.07 ± 0.01	1.20 ± 0.08	1.79 ± 0.04	1.92 ± 0.14
0.71	0.07 ± 0.01	1.06 ± 0.07	1.82 ± 0.09	1.93 ± 0.13

Although the results indicate a change in alloy constitution from the as-received material to the condition prior to rolling pass 2 (or, equivalently, after rolling pass 1 and annealing for 35 s), the analysis planes providing the coarse and fine grained regions show practically no difference in second-phase particle parameters within the confidence limits of the measurements. Furthermore, limited TEM studies and electrical conductivity measurements indicate that within the present processing times no second phase particle instabilities were observed in the samples mentioned

above [7]. Therefore, the measured changes in fraction recrystallised and recrystallised grain size are expected to essentially relate to changes in original grain size only.

Fig. 1 shows the volume fraction measurements as a function of annealing time for the various original grain sizes. The smooth curves have been obtained by non-linear regression analysis on the JMAK equation:

$$X = 1 - \exp\left[-0.693(t/t_{0.5})^k\right] \tag{3}$$

where X is the volume fraction recrystallised in time t and k the Avrami exponent. Notice that as the original grain size increases, the recrystallisation kinetics are slower.

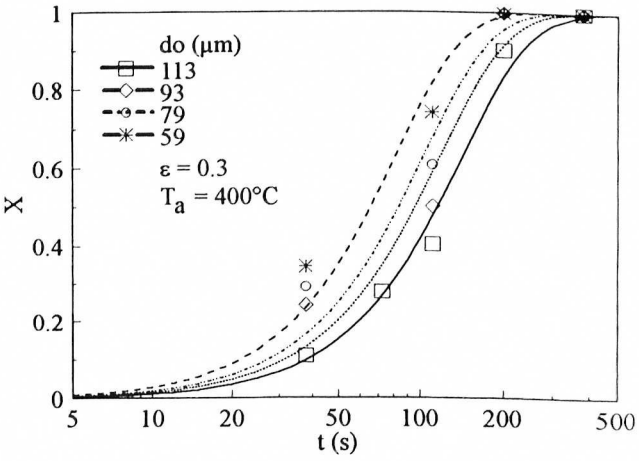

Fig. 1. Fraction recrystallised as a function of annealing time for samples with different original grain sizes.

On the other hand, Fig. 2 and Fig. 3 show $t_{0.5}$ and d_{rex} as a function of original grain size, respectively, with data from a similar Al-1Mg alloy tested in PSC at 400°C [2]. Equation (1) and (2) were used in order to fit the experimental data. Thus, non-linear regression analysis gave $a = 0.88$ and $b = 0.73$ for $\varepsilon = 0.3$, but $b = 0.46$ for $\varepsilon = 0.71$, exponents which are lower than those obtained by Puchi [2], namely, $a = 1.35$ and $b = 1.3$ for $\varepsilon = 0.71$. Interestingly, as the strain increases the effect of original grain size becomes increasingly smaller. $b = 0.73$ for $\varepsilon = 0.3$ implies grain corners and edges are dominant, but $b = 0.46$ for $\varepsilon = 0.71$ indicates edges and surfaces prevail. It thus suggests that there is a change in preferential nucleation sites, which can be interpreted as follows. The total nuclei density can be expressed as [8]

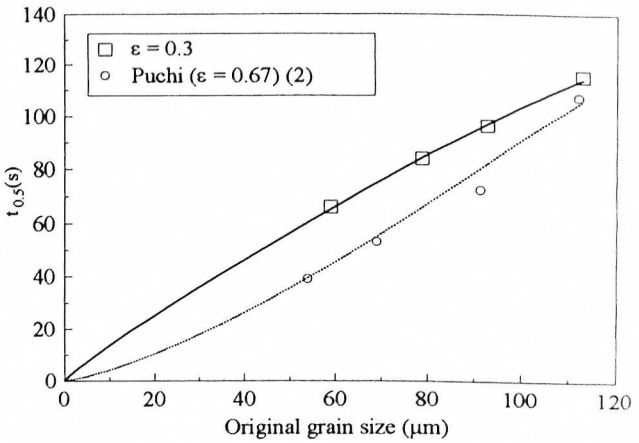

Fig. 2. $t_{0.5}$ as a function of d_0.

$$N_v = N_v^c + N_v^e + N_v^s + N_v^i \tag{4}$$

where N_v^c, N_v^e, N_v^s and N_v^i are the nuclei density contributed by grain corners, grain edges, grain surfaces and intragranular sites, respectively. It is worth mentioning that an extra term could be added to equation (4) in order to take into account particle stimulated nucleation. At low strains, corners should dominate, with a transition to edges and surfaces as the strain increases, whereas high strains should lead to significant intragranular nucleation. As $N_v^c \propto d_0^{-3}$, $N_v^e \propto d_0^{-2}$, $N_v^s \propto d_0^{-1}$ and N_v^i is independent of d_0 and, furthermore, $d_{rex} \propto N_v^{-1/3}$, then a decaying effect of d_0 on d_{rex} is consistent with such a transition. This is in agreement with recent findings [9] in the same alloy; a lamellar structure is observed inside the deformed grains comprising boundaries forming microbands and random sub-boundaries within the microbands. Such a structure provides intragranular sites as it has been found that the area of these boundaries is much greater than the "old" grain boundary area, leading to a high probability of successful nucleation inside the deformed grains. In fact, the misorientation continues to increase with increasing strain within the steady state deformation regime across the lamellae boundaries.

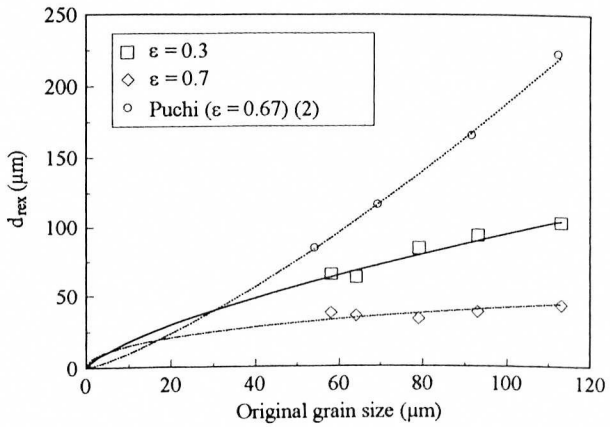

Fig. 3. d_{rex} as a function of d_0.

Conclusions

The experimental results indicate that the original grain size has an important effect on static recrystallisation, but the effect becomes increasingly small as the strain increases. Grain boundary regions are important nucleation sites at low strains, but high strains activate intragranular sites. A complete nucleation model should be able to predict such a transition.

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