RECRYSTALLISATION BEHAVIOUR OF AN Al-1Mg ALLOY DURING THERMOMECHANICAL PROCESSING

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Abstract

Static recrystallisation, which may occur between deformation passes under hot working conditions, is important for the microstructural evolution of aluminium alloys. It is characterised by the kinetics, the recrystallised grain size and texture developed. The current interest in modelling hot rolling demands quantitative relationships between the microstructural parameters, the processing variables (strain, strain rate, temperature and time) and the prior thermomechanical history of the stock. Since basic physical metallurgy does not provide these relationships as yet, one has to resort to empirical - physically sound - equations. This paper presents these relationships for a characteristic recrystallisation time, the recrystallised grain size and the size of the recrystallising grains as a function of processing variables obtained by quantitative optical metallography after hot rolling and annealing of tapered slabs of Al-1Mg. The predictions are compared with available data from plane strain compression tests of a similar alloy. An insight into the nucleation kinetics and growth rate variations is also provided.

Introduction

The kinetics of isothermal static recrystallisation usually follows closely an Avrami equation

$$X_v = 1 - \exp\left[-0.693 \cdot \left(\frac{t}{t_{0.5}}\right)^k\right] \qquad (1)$$

where $t_{0.5}$ is the time at which 50% of the material has recrystallised. Following the observed behaviour in steels,(1-6) several workers have developed empirical equations for $t_{0.5}$. Plane strain compression tests of two casts of Al-1Mg led to (5,6)

$$t_{0.5} = \phi \cdot d_o^{1.35} \cdot \epsilon^{-2.7} \cdot Z^{-1.1} \cdot \exp\left(\frac{230000}{R \cdot T_a}\right) \qquad (2)$$

where ϕ is a constant (9.6×10^{-6} and $3.6 \times 10^{-6} \mu m^{-1.35} \cdot s^{2.1}$ for cast A and cast B, respectively), d_o the original grain size, ϵ the equivalent strain, Z the Zener-Hollomon parameter with an activation energy for deformation of 156 $kJmol^{-1}$, R the gas constant and T_a the annealing temperature.

A similar expression has been used for the recrystallised grain size (1-6), which for plane strain compression of Al-1Mg was reported to be (5,6)

$$d_{rex} = \psi \cdot d_o^{1.3} \cdot \epsilon^{-0.39} \cdot Z^{-0.24} \qquad (3)$$

with no annealing temperature dependence. The constant ψ was 435 and 185 $\mu m^{-0.3} \cdot s^{0.24}$ Grain coarsening, expected after recrystallisation is complete, is usually negligible in the times of interest in hot rolling (6).

During recrystallisation, the size of the recrystallising grains d_r is given by (3)

$$d_r = X_v^{1/3} \cdot d_{rex} \qquad (4)$$

if nucleation is site saturated. This relationship seems to be closely followed after relatively high strains (0.7-5) in hot axisymmetric compression and torsion (3). In addition, an equation based on site saturation has been used for modelling the mean grain size of hot rolled aluminium alloys (7).

Experimental Procedure

The chemical composition of the alloy considered is given in Table I. It is actually an AA1050 alloy with 1wt%Mg. After DC casting, the alloy was homogenised and hot rolled to produce a fully recrystallised structure of nearly equiaxed grains, $178 \pm 6 \mu m$ in mean size with an aspect ratio of 1.5 ± 0.1, containing low volume fraction of Mg_2Si ($< 1.5 \mu m$), AlFeSi ($0.25 \mu m$) and eutectic ($2 \mu m$) particles.

Table I. Chemical Composition

Al	Mg	Fe	Si	Mn	Others
98.56	0.94	0.305	0.147	0.001	0.047

Tapered slabs (Figure 1a) were selected for the rolling tests because they present some advantages for double deformation tests and also because they provide a range of processing conditions to examine in a single slab. A fully instrumented 50 tonne rolling mill was used together with two electrically-heated muffle type furnaces situated to either side of the mill. The slabs were reheated to 395, 455 and 510°C and rolled to 12 mm thickness ($0.20 \leq \epsilon \leq 0.71$ and $3.6 \leq \dot{\epsilon} \leq 5.9 s^{-1}$) at 385, 445 and 505°C, with subsequent annealing for different times at 350, 400 and 450°C, respectively. After annealing, the slabs were quenched, and samples 26x12x5 mm (Figure 1b) were prepared metallographically (grinding, polishing and anodising) for measurements under cross-polarisers.

The quantitative metallography was performed as follows. A base direction was selected, the width direction in this case, and the sample was traversed twice at 3.5 mm from the surface; the first time to measure volume fraction recrystallised by point counting; the second time to count three type of boundaries per unit length: (i) between unrecrystallised grains, (ii) between recrystallised grains and (iii) between recrystallised and unrecrystallised grains. The second traverse provides the size of the recrystallising grains as well as the migrating boundary area per unit volume.

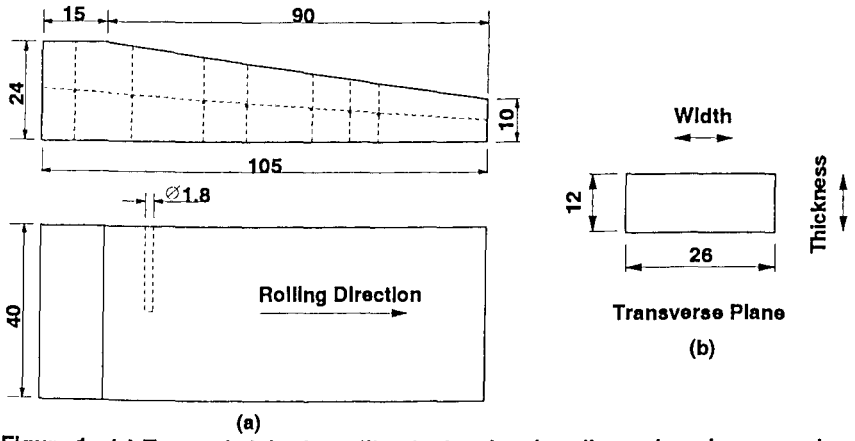

Figure 1 . (a) Tapered slabs for rolling tests, showing dimensions in mm and location of thermocouple, (b) Sample dimensions and orientation.

Results and Discussion

It has been found that recrystallisation closely follows the Avrami kinetics with $0.8 < k < 1.8$ k increases with increasing strain but shows no systematic temperature dependence. Furthermore, recrystallisation occurs without hindrance from fine second-phase particles, and practically no grain coarsening takes place within the present processing times. Figure 2 presents the results for the kinetics which have been fitted as

$$t_{0.5} = 5.3x10^{-13} \cdot \epsilon^{-1.6} \cdot Z^{-0.2} \cdot \exp\left(\frac{205000}{R \cdot T_a}\right) \qquad (5)$$

The activation energy for deformation was assumed to be 156 $kJmol^{-1}$. The activation energy for recrystallisation of 205 $kJmol^{-1}$ was obtained from separate experiments in which samples from a single rolled slab were annealed at different temperatures. Eq.(5) gives a reasonable fit to the experimental data for $0.28 \leq \epsilon \leq 0.71$. The data corresponding to $\epsilon = 0.2$ are away from predictions using Eq.(5), disclosing a deficiency of this equation. Taking into account that $t_{0.5}$ relates to flow stress through dislocation density, and that flow stress under hot working conditions is fitted by a Voce equation (6)

$$\sigma = \sigma_o + (\sigma_s - \sigma_o) \cdot [1 - \exp(-C \cdot \epsilon)]^m \qquad (6)$$

where σ_o and σ_s are initial and steady state flow stresses, respectively, $m \approx 0.5$ and C is a constant dependent on the strain to the onset of steady state; the power laws for ϵ and Z are not strictly correct. Figure 3 compares the experimental results with calculated values using the same processing parameters and an original grain size of 178 μm in Eq.(2). The present alloy recrystallises faster than cast A and cast B by a factor of about 8-22 and 3-8, respectively. The discrepancy with experimental values is somewhat lower at 350°C but much higher at 450°C.

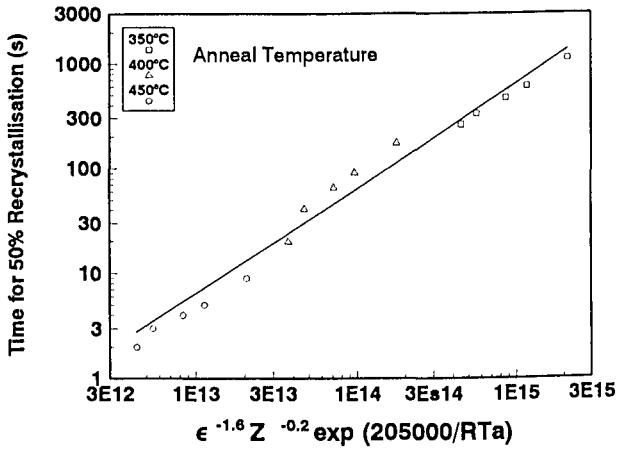

Figure 2 . Recrystallisation kinetics after one deformation. Original grain size: 178 μm.

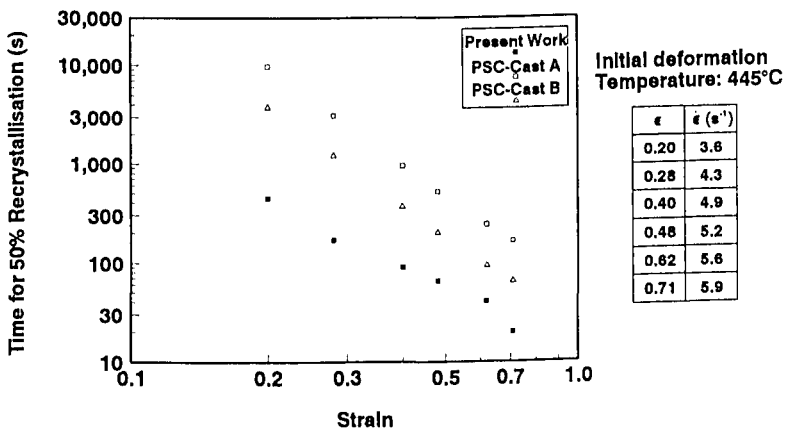

Figure 3 . Recrystallisation kinetics. Comparison with prediction from Eq.(2) at 400°C and strain rates corresponding to indicated strains.

Figure 4 shows the recrystallised grain size obtained in the present work, which has been fitted as

$$d_{rex} = 1350 \cdot \epsilon^{-0.85} \cdot Z^{-0.1} \qquad (7)$$

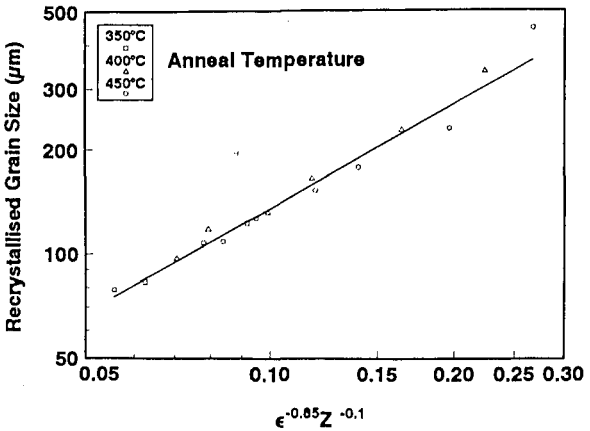

Figure 4 . Recrystallised grain size after one deformation. Original grain size: 178 μm.

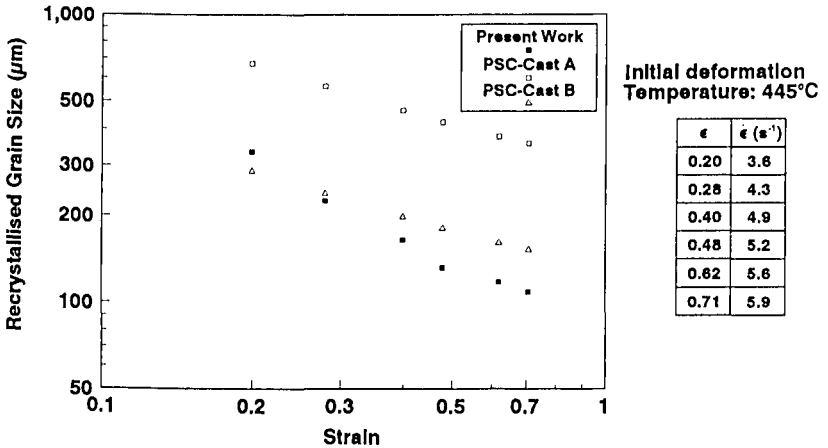

Figure 5. Recrystallised grain size. Comparison with prediction from Eq.(3) at 400°C and strain rates corresponding to indicated strains.

Eq.(7) seems adequate for the present processing conditions. Figure 5, on the other hand, compares the data with predicted results from plane strain compression tests of the same two casts as for the kinetics. The grain size in cast B is much closer to the present results (the factors are 2-3 and 1-1.4 for cast A and B, respectively). A similar picture is observed at 350 and 450°C. Clearly there is a difference in both kinetics and recrystallised grain size, partly due to the difference in exponents and activation energy as well as constants in the equations, but the inconsistency is considerably reduced if the grain size dependence in Eq.(2) and (3) is considered to be too high (5). There is evidence that alloy constitution differences, such as

produced by different casting and homogenisation conditions have a great effect on both kinetics and recrystallised grain size (5). Some contribution due to the different deformation modes is also to be expected although it is not possible to separate the effects in this case.

The recrystallising grain size is shown in Figure 6, compared with predictions using Eq.(4) for different strains. Note that Eq.(5) does not fit the present results indicating that site saturation does not hold. New small grains continue to nucleate as recrystallisation proceeds.

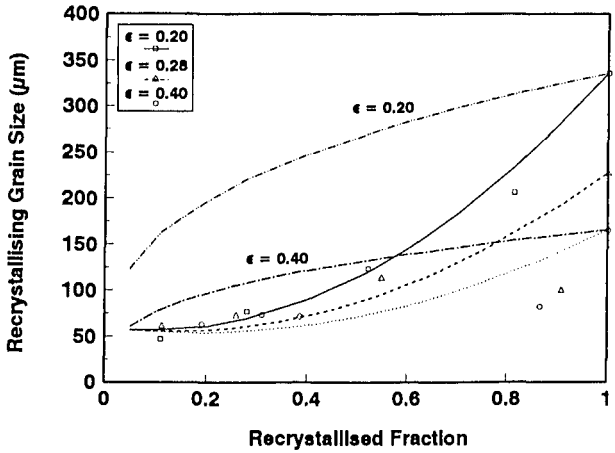

Figure 6 . Recrystallising grain size after one deformation. Chain lines were obtained by using Eq.(4) at strains of 0.2 and 0.4, respectively.

The procedure developed by Gokhale and DeHoff (8) and employed by Vandermeer and Rath (9-10) to determine nucleation rate, which assumes random nucleation, gave

$$\dot{N}_v = 11.8 \cdot t^{-0.41} \qquad (8)$$

with the nucleation rate $\dot{N}_v$ in $mm^{-3} \cdot s^{-1}$, and the anneal time in s, for a strain of 0.28 at 400°C. Eq.(8) shows a continuously decreasing nucleation rate with increasing time. If it is assumed that the recrystallising grains are tetrakaidecahedra of uniform size, it can be shown by using basic stereological relationships that (11)

$$N_v^r = 0.43 \cdot X_v \cdot d_r^{-3} \qquad (9)$$

Figure 7 shows measured number of recrystallising grains per unit volume at different anneal times, using Eq.(9). This Figure also shows N_v^r obtained by integration of Eq.(8) for comparison. Although the number of data points is limited, it is clear that N_v^r is not constant during recrystallisation.

Growth rates variations during recrystallisation have also been observed, and Figure 8 presents the results for three different strains at 400°C. These variations are thought to be due to either concurrent recovery and/or the deformation zones formed around large undeformable particles (12). In the present work, the largest particles are about 2 μm in size which is below the critical size to form deformation zones (13). However, longer range heterogeneities in stored energy are still expected.

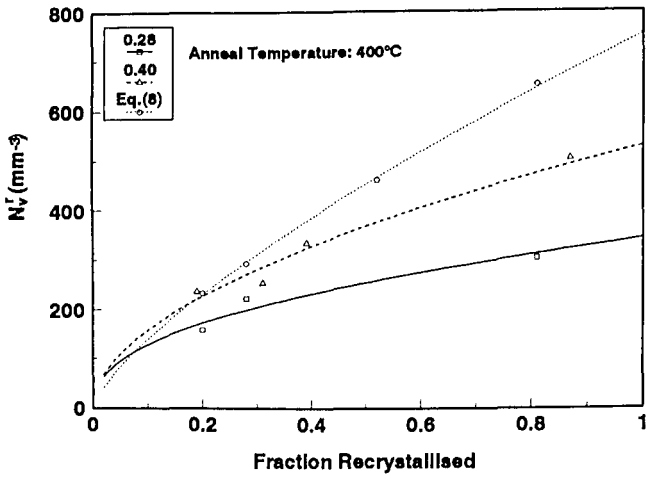

Figure 7 . Recrystallising grains per unit volume at different annealing times.

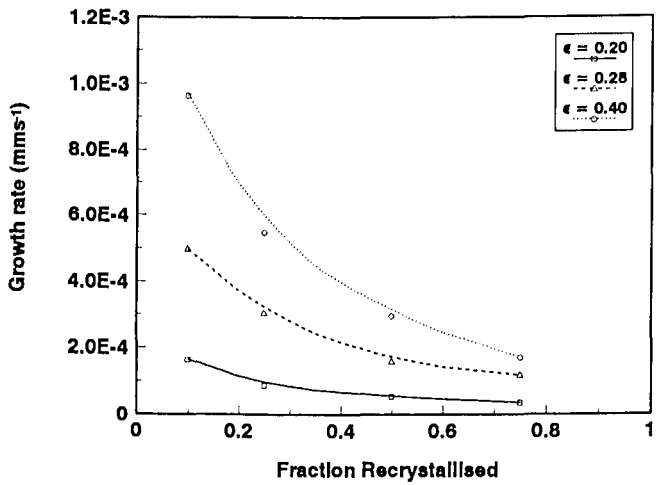

Figure 8 . Growth rate variations during recrystallisation.

Conclusions

Empirical equations for $t_{0.5}$ and d_{rex} as a function of processing parameters have been developed for hot rolled Al-1Mg alloy. The observed discrepancies with derived PSC tests of similar alloys seem to be a result of difference in alloy constitution and possibly in deformation mode. Further work is required aiming at a better equation for $t_{0.5}$ by means of a physical model. Although the appropriate form of equation for d_r is still uncertain, site saturation does not hold at strains in the range $0.2 \leq \epsilon \leq 0.4$. Instead, nucleation appears to occur continuously at decreasing rates during recrystallisation. Growth rate variations also take place.

Acknowledgments

One of the authors currently undertaking his PhD at the University of Sheffield, P.L. Orsetti Rossi, acknowledges C.V.G. Industria Integrada de Aluminio (Bauxilum) C.A., the University of Sheffield and the Overseas Research Scheme for their sponsorship. The authors also thank Alcan International, Banbury, for the supply of the Al-1Mg alloy.

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