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Modeling the effect of Al_3Sc precipitates on the yield stress and work hardening of an Al–Mg–Sc alloy

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Abstract

The yield stress and work hardening behaviour of precipitation hardening alloys is directly related to the nature of the interaction between mobile dislocations and precipitates. In commercial systems such as aluminium alloys, the understanding of this problem is complicated by the overlap between various mechanisms and the interplay between the volume fraction and size of precipitates and the residual solid solution content. In this study a model Al–2.8 wt.% Mg–0.16 wt.% Sc alloy has been chosen for examination since precipitation involves simple spherical precipitates, the absence of metastable phases and the solid solution effect is dominated by the magnesium content. Using a previously development precipitation model, it has been possible to develop an integrated yield stress/work hardening model in which the shearable/non-shearable transition and the size distribution of precipitates are explicitly accounted for. The agreement between the model and the experiments is excellent and the shearable/non-shearable transition radius is consistent with experimental observations.

Keywords: Aluminium alloys; Ageing; Plastic deformation; Yield phenomena

1. Introduction

Over the past 50 years, considerable progress has been made in developing a framework for modeling the strength and work hardening of face-centred cubic (fcc) and body-centred cubic (bcc) metals. It is generally accepted that strength arises from dislocation–dislocation interactions and work hardening represents a competition between storage of dislocations which is geometric in nature and thus athermal, and the loss of dislocations by dynamic recovery processes which are temperature and solute dependent [1,2]. At the same time, studies on strengthening mechanisms for precipitation hardening alloys have focused on understanding the yield stress in terms of dislocation–precipitate interactions [3–6]. However, for alloys of commercial interest, multiple dislocation–precipitate interaction mechanisms often overlap making the development of detailed models challenging. As such, recent yield stress models have assumed a generic linear relationship between precipitate strength and size for precipitates that are sheared by moving dislocations [7–9]. The radius marking the transition from dislocation shearing of the precipitates to dislocation bypassing then assumes the critical adjustable parameter in models which capture a physical view of the age hardening curve.

The yield stress represents only one aspect of the plastic response for precipitation hardening alloys. It has been well known since the seminal study of Byrne and co-workers [10] that the work hardening behaviour of copper single crystals changed dramatically as the material was aged. Recently, Poole and co-workers have examined the systematic changes in work hardening of 6000 and 7000 series aluminium alloys for a range of ageing conditions [11–13]. The results are complex with the initial work hardening rate first decreasing and then increasing, and the apparent rate of dynamic recovery showing a sharp increase corresponding to the shearable/non-shearble transition. While initial

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attempts have been made to understand these effects, many questions still exist. The challenge remains to develop a comprehensive framework for modeling yield stress and work hardening that can encompass the complexity found in commercial alloys.

If we return to our basic framework, one can consider the possible mechanisms by which plastic deformation is modified in precipitation hardening alloys. First, the storage of dislocations is affected by the presence of precipitates. For shearable precipitates, evidence exists in some alloy systems for localization of slip due to local softening on the slip plane as the precipitate cross-section is reduced by the shearing process [14,15]. On the other hand, for the case of precipitate bypassing, dislocation loops are stored around the particles. These loops may rearrange through various relaxation mechanisms as deformation proceeds, leading to an increased global storage rate for dislocations [16,17]. Alternatively, one must consider the role of elastic stresses borne by the particles [18,19], particularly for cases where the volume fraction is several percent and the shape of the particles is efficient for load transfer, i.e. the case of plates, rods or laths [20]. Second, dynamic recovery can be affected as the matrix solute level decreases as precipitation proceeds [20] and it is possible to envision cooperative mechanisms for dislocation annihilation when closely spaced, non-shearable precipitates are present (noting that the storage of dislocations in adjacent precipitates is redundant) [12]. Finally, careful consideration must be given to the question of how to add flow stress contributions when mobile dislocations interact with multiple obstacles on the glide plane (forest dislocations, solute atoms and precipitates) [3,6,21,22].

This complexity points to the need for the study of model alloy systems where systematic experiments can be conducted to separate the various possible mechanisms. With this in the mind, the Al–Mg–Sc system was chosen as a model system with the following advantages: (i) the precipitation sequence is simple without the complexity of metastable transition phases [23]; (ii) the solid solution effect on work hardening will be dominated by the magnesium, i.e. the small changes in the scandium solid solution during precipitation are negligible compared to the base magnesium level and (iii) due to a low volume fraction of spherical precipitates (i.e. inefficient for load transfer) it is expected that the elastic stresses borne by these particles will be negligible. In addition, it is worth noting that magnesium is almost insoluble in Al_3Sc [24–26] such that the precipitation kinetics and the hardening effect of Al_3Sc precipitates resembles that of binary Al–Sc alloys where the strengthening effect of Al_3Sc has been experimentally characterized in a number of studies, see Refs. [24,26–31].

In summary, the main objective of the present work is to develop a self-consistent model for the yield stress and work hardening behaviour of an Al–Mg–Sc alloy over a range of ageing conditions progressing from shearable to non-shearable precipitates. The precipitate size distribution will explicitly be considered using a precipitation model presented in a companion paper [32].

2. Experiments

The material for this study was produced in a laboratory scale DC casting apparatus at the Novelis Global Technology Centre (Kingston, ON). The composition of the alloy is shown in Table 1. Plates with a thickness of 5 mm were cut from the ingots and then homogenized at 610 °C for 7 days in a box furnace. After 7 days the material was removed from the furnace and water quenched to room temperature. This material was cold-rolled to 1 mm in thickness and then tensile samples punched out using a die. The thermal treatments were conducted with a salt-bath having a composition of 40% $NaNO_3$ + 60% KNO_3. In order to set the starting grain size, the cold-rolled material was recrystallized using a short anneal of 5 s in the salt-bath at 610 °C followed by water quenching. The grain size following the flash anneal was found to be 45 μm. After this treatment, samples were subjected to isothermal heat treatments at 300 and 350 °C. Two sets of up-quenching experiments were also performed to allow for rapid coarsening of the precipitates. In the first of these coarsening experiments samples were aged at 300 °C for 510 min followed by ageing at 425 °C. The second up-quenching experiment involved ageing at 300 °C for 510 min followed by ageing at 350 °C. In all cases, samples were water quenched following annealing.

Mechanical behaviour has been characterized in terms of uniaxial tensile tests conducted with samples immersed in liquid nitrogen (77 K) to avoid the effects of dynamic strain ageing observed at room temperature. Tests were conducted with a 5 kN load cell, at a nominal strain rate of 10^{-3} s^{-1} with an extensometer attached to the sample. The work hardening rate was determined by numerically differentiating the true stress with respect to true strain followed by smoothing using polynomial fitting to successive segments of the curve.

The spatial arrangement of dislocations after plastic deformation was investigated using transmission electron microscopy (TEM). Thin foils were prepared in the conventional manner by mechanical polishing, punching and jet polishing using an electrolyte consisting of 5% percholoric acid in methanol operating at −20 °C at 20 V. The foils

Table 1
Chemical composition of the alloy (wt.%)

Mg	Sc	Mn	Fe	Ti	Si	Cr	Cu	Ni	V	Zn	Zr
2.90	0.16	0.001	0.05	0.13	0.03	0.001	0.001	0.001	0.009	0.001	0.001