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Diffusion creep and superplasticity in aluminium alloys

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Abstract

The plastic deformation of two classes of fine-grained aluminium alloys at elevated temperatures and slow strain rates have been investigated. One class of material, Al–Cu–Zr, was processed to develop banded microstructures; the other class, based on Al–(Mg)–Mn, had near-equiaxed microstructures. In both classes, superplastic behaviour was found in the variants with the higher solute content. The evolution of the banded microstructures and the results from surface grid measurement in the Al–(Mg)–Mn alloys give results which indicate that the superplasticity is primarily a result of diffusion creep, and the effect of solute is proposed to be via an enhancement of solvent self-diffusion.

Keywords: Superplasticity; Aluminium; Diffusion creep

1. Introduction

The occurrence of very large elongations, associated with an unusually high strain rate sensitivity of flow stress, in fine-grained polycrystalline materials – especially metals – has been known for many years. It is referred to as superplasticity and has been the subject of much research. Despite the amount of effort devoted to the phenomenon, there remain uncertainties about the fundamental mechanism responsible for the high rate sensitivity.

It is clear that a fine grain size is required for the high elongations. A much simplified view of the microstructural development is that equiaxed grains remain equiaxed even after large deformations, and this – together with some other observations including the study of surface phenomena – has led to the commonly held view that grain boundary sliding (GBS) is the fundamental microstructural mechanism of superplasticity, even if other, accommodation, mechanisms are often invoked to try to explain the high level of strain rate sensitivity of flow stress. This mechanism is more correctly referred to as Rachinger sliding, where the grains move relative to each other with little intragranular deformation [1]. Despite the popularity of this view, there are observations which are not consistent with the operation of Rachinger GBS. Important examples of these are found in a class of superplastic aluminium alloys.

Superplastic deformation in aluminium has been exploited commercially for a number of years. Although two-phase alloys, generally based on eutectic compositions, have been investigated [2–4], attention has been focused on alloys with only a small volume fraction of second phase. That second phase is used to provide a dispersion of small particles which control grain growth via the Zener pinning mechanism. Two distinct methods of producing the fine grain size required have been developed. In the first of these, sheet is produced using hot rolling in material cast and heat treated to contain fine $ZrAl_3$ particles. The temperature during hot rolling leads to significant static and dynamic recovery, and discontinuous recrystallisation does not occur. The microstructure of these materials at the start of superplastic deformation comprises bands (ribbons) of similarly oriented subgrains. During superplastic deformation, the misorientations between the subgrains increase and eventually a microstructure of small grains separated

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by high-angle boundaries develops. Examples of this type of alloy are the SUPRAL Al–Cu–Zr type [5,6] and AA8090 Al–Li–Cu–Mg–Zr. In the other class of materials, discontinuous recrystallisation occurs after rolling and prior to superplastic forming, giving a microstructure consisting of approximately equiaxed grains prior to superplastic deformation.

There has been a significant amount of research on the materials with banded microstructures. In the conventional Rachinger GBS model, the initial microstructure would not be expected to be superplastic because of the disposition of high-angle grain boundaries relative to the straining system. Ash and Hamilton [7] gave results which indicated that high rate sensitivity was only found after deformation had led to a break-up of the banded structure, though this was not found in other studies [8,9], where the rate sensitivity was found to be almost constant with straining. There is also a question of whether non-superplastic deformation would lead to the break-up of the banded structure – which occurs by grain/subgrain rotation, giving orientation divergence. Although this effect occurs, it is still possible to observe orientation correlation in bands after quite large superplastic strain, and this use of the bands as "internal markers" gives results which are not consistent with Rachinger GBS [8,10].

There has been a long history in the use of surface markers to study superplastic deformation at the microscale. Most of this has relied on scratch marks, which are only of value in showing the development of grain boundary offsets. Photolithographic grids have been used in the study of creep [11], and more recently focused ion beam (FIB) micromilling has been used to produce grids of controlled size. These show that intragranular strains are very small in superplastic deformation [12]. This would be consistent with Rachinger GBS.

There is, then, a conflict between the persistence of orientation banding, which appears contrary to Rachinger GBS, and the microgrid results, which seem to support that type of mechanism. In this study, a set of aluminium alloys are used which not only allow further study of that apparent contradiction, but also use compositional variables – the level of solute – to help with investigation of the mechanism of superplasticity.

2. Experimental work

2.1. Materials

Four alloys were used in this study, with the compositions given in Table 1. Two of them were based on Al–Cu–Zr, were produced using high-purity melt stock and were processed to give the "banded" microstructure typical of commercial SUPRAL and AA8090 alloys. The other two were commercially produced alloys based on Al–Mn, with and without a significant Mg content. These were processed to give discontinuously recrystallized microstructures.

The Al–Cu–Zr alloys were received as DC cast billet, with a high degree of superheat in the casting to ensure that the dissolved zirconium content was high. The cast material was heat treated at 350 °C for 16 h to precipitate fine $ZrAl_3$ before holding at 500 °C for 2 h and water quenching to dissolve the copper. The materials were then hot rolled, with an initial temperature of 300 °C and periodic re-heating, to 1.5 mm thickness. The different copper contents gave rise to different grain sizes after casting and heat treatment – the low copper alloy had a grain size more than twice that of the high copper material – and to produce sheet of approximately the same thickness and with similar microstructures, the initial thickness of the Al–2Cu–Zr section used for rolling was 50 mm, and for the Al–4Cu–Zr it was 23 mm. The sheets were then annealed at 450 °C for 1 h. The microstructures are shown in Fig. 1. These orientation images, and others given below, were obtained using a FEI Sirion FEG-SEM with HKL Channel 5 electron back-scattered diffraction (EBSD) acquisition software, and used a step size of 0.5 μm. The data were plotted and analysed using VMAP software [13].

Table 1
The compositions (in wt.%) of the alloys used.

Alloy	Cu	Mg	Mn	Zr	Si	Fe	Cr	Others
Al–2Cu–Zr	**2.01**	0.001	0.001	**0.29**	0.006	<0.001	0.001	<0.02
Al–4Cu–Zr	**3.81**	0.001	0.001	**0.39**	0.021	0.002	0.001	<0.02
AA3003	0.10	**0.00**	**1.00**	0.00	0.14	0.58	0.00	<0.02
AA5083	0.10	**4.67**	**0.70**	0.00	0.02	0.03	0.11	<0.01

The Al–(Mg)–Mn alloys relied on Al_6(Mn, Cr) dispersoids to give control of the grain size via Zener pinning. The Al–Mg–Mn alloy AA5083 was received as cold-rolled sheet. This was annealed at 530 °C for 2 h to give a fine, recrystallized microstructure, as shown in Fig. 2b. The Al–Mn alloy AA3003 was received as a 41 mm thick slice from a commercial DC cast billet. This was annealed at 600 °C for 3 h and water quenched before hot rolling at 400 °C to 31 mm and then cold rolling to a thickness of ~1.8 mm. This sheet was annealed at 530 °C for 1 h, giving the microstructure shown in Fig. 2a.

2.2. Tensile testing

All tensile tests were carried out on specimens with 12.7 mm parallel gauge section 6.3 mm wide and having simple 20 mm wide tag ends with no blend radii. The tensile axis in all cases was parallel to the rolling direction (RD) of the sheet. Tests were carried out in air, and the specimens were held at the test temperature −450 °C for the Al–Cu–Zr alloys and 530 °C for the Al–(Mg)–Mn alloys – for 20 min prior to straining. Further details of the testing configuration are given elsewhere [14]. Tests were carried out with either a constant strain rate or with the strain rate switching between two values. The perturbed strain rate gave a stepped response in stress, which allows the rate sensitivity index, m, to be determined [9]. In deriving stress–strain curves, the gauge length was assumed to have been parallel and of constant volume throughout. It is known that neither of these assumptions is actually true [14], and for microstructural results where