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Evolution of Microstructure and Texture during Hot Deformation of a Commercially Processed Supral100

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The microstructure and texture in a commercially processed Al-6 wt% Cu-0.4 wt% Zr (Supral100) aluminium alloy have been investigated after annealing and hot tensile straining at 450 °C, using a field emission gun scanning electron microscope (SEM) and electron backscatter diffraction (EBSD). The microstructure of commercially processed alloy had a relatively large fraction of high angle grain boundaries (HAGBs) which were aligned parallel to the rolling direction, and a strong texture. Annealing at 450 °C led to an increase in the fraction of HAGBs and to an increase in HAGB spacing and these changes were progressively enhanced by subsequent tensile deformation. The increasing fraction of HAGBs was due to the annihilation of low angle grain boundaries (LAGBs). A sharpening of texture during annealing was attributed to preferential textural growth, and the reduction of texture at higher tensile strains led to the development of superplastic behaviour. The present work supports the view that the evolution of the fine grain microstructure during the high temperature straining of Supral100 is primarily due to the accumulation of a large area of grain boundary during the initial thermomechanical processing, and does not involve any unusual restoration processes.

KEY WORDS: **Aluminium; Superplastic deformation; Microstructure; Texture; Electron backscatter diffraction**

1. Introduction

For polycrystalline solids to exhibit superplastic flow they must have equiaxed fine grain (<10 μm) microstructures, comprised mainly of high angle boundaries, which are stable at relatively high testing temperatures (>0.5 T_m, where T_m is the melting point in degrees Kelvin). For commercial Al alloys processed by the ingot route, the required microstructures are produced after thermomechanical processing (TMP) either by static recrystallisation prior to superplastic forming (SPF), or microstructural evolution during the early stages of SPF[1]. The former procedure is used for alloys such as AA5083 (Al-Mg-Mn) and AA7475 (Al-Zn-Mg-Cr), whereas AA2004 (Supral100) is made superplastic normally by the latter route. Recently, efforts have been paid to enhance the superplastic performance of the material by severe plastic deformation, which allows submicron grain structures to be developed in the material[2], although this route is expensive and applicable to limited small scales.

Supral100 has a nominal composition in wt% of Al-6Cu-0.4Zr, and consists of an Al-Cu solid solution which contains a dispersion of very fine (<10 nm) $ZrAl_3$ precipitates and some $CuAl_2$ particles. During production the material is cast from a high superheat (∼780 °C), and rapidly cooled to avoid the formation of coarse $ZrAl_3$ particles and obtain a high level of Zr in solid solution. The material is aged at 360 °C to precipitate fine $ZrAl_3$ particles, solution treated at ∼50 °C to take most of the copper into solid solution, and hot rolled to break down the as-cast structure. A subsequent warm/cold rolling schedule produces a heavily cold worked structure from which a fine microstructure evolves during the early stages of SPF at ∼450 °C.

After TMP the material is stabilised against static

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recrystallisation by $ZrAl_3$ dispersoids, and the evolution of a fine grain superplastic microstructure with strain at 450 °C is attributed to continuous dynamic recrystallisation. However, this phenomenological description does not imply the operation of any specific mechanism. It is usually assumed that after processing the microstructure consists of elongated, highly dislocated, pre-existing grains from which subgrains of low misorientation develop during the early stages of elevated temperature deformation or on annealing. Humphreys and Hatherly[3] have described mechanisms by which recrystallisation may occur dynamically. These include subgrain growth, subgrain coalescence, subgrain rotation or assimilation (accumulation) of dislocations into subgrain boundaries[4–8]. Ridley *et al.*[9] studied Al-Cu-Zr alloys and proposed that the structural evolution observed was consistent with strain induced geometrical dynamic recrystallisation. McNelley *et al.*[10,11] characterised the microstructure of Supral100 and reported that after TMP the material contained deformation bands of alternating lattice orientation, parallel to the rolling direction, which corresponded to the symmetric variants of the brass texture component {011}<211>. During annealing at 450 °C, high angle grain boundaries (HAGBs) develop from the transition regions between the bands, and it was proposed that these contribute to the development of small equiaxed grains during subsequent hot/superplastic deformation.

In favour of the TMP plus hot deformation route for exploiting the superplasticity of the material, which is cheaper, more efficient and commercially viable than other alternative routes such as severe deformation processing, the present work describes a study which examines microstructural evolution during hot deformation of commercially processed Supral100 sheet, in order to clarify the mechanisms leading to the establishment of superplastic behaviour in the material.

2. Experimental

The alloy investigated in the experiment was Supral100 of nominal composition Al-6 wt%Cu-0.4 wt%Zr supplied in the form of sheet by Superform, Worcester. The initial sheet thickness was 1.6 mm and the cast and heat treated alloy had been processed by a route which involved warm/cold rolling.

Prior to tensile straining, the effect of annealing at various temperatures up to 450 °C, for 1 h, on microstructure and texture was studied. Tensile specimens with a gauge length of 22 mm were machined from the sheet material with the tensile axis parallel to the rolling direction, and annealed at 450 °C for 1 h to remove any mechanical damage. The specimens were deformed at 450 °C at a constant strain rate of 10^{-3} $s^{-1}$to predetermined strains using an Instron tensile machine fitted with a high temperature testing facility. After deformation the specimens were cooled rapidly to ambient temperature. They were then sectioned longitudinally parallel to the RD-ND plane through the centre of the specimen and, after mechanical polishing and electropolishing, were examined metallographically. Additional measurements were also made on sections parallel to the RD-TD orientation.

Detailed characterisation of the samples was made in a Camscan Maxim 2040 FEGSEM, by both backscattered and secondary electron imaging and electron backscatter diffraction (EBSD). EBSD maps were obtained using an HKL Channel acquisition system and processed using Vmap, an in-house software development. The positions of HAGBs (>15°) and low angle grain boundaries LAGBs) (1°–15°) in the EBSD maps were determined, the lower angular limit being selected to remove noise from the data as discussed by Humphreys[12]. The maps presented in this paper show the HAGBs as black lines. The LAGBs are not shown for clarity. The amounts of various identified texture components were determined from the EBSD maps, and a map pixel was deemed to have the specified texture if its orientation was within 15° of the ideal.

3. Results and Discussion

3.1 As-received material

Fig. 1 is an EBSD map of the as-received microstructure of the as-received bulk material in the RD-ND plane, for the 1.6 mm thick sheet. Regions containing larger $CuAl_2$ particles are filled black, and for clarity, only the HAGBs are marked. A highly elongated or banded structure aligned parallel to the rolling direction is seen and is consistent with the heavy deformation that the material has received. It may be noted that the distribution of the large second phase particles and of the grain boundaries shows some heterogeneity. EBSD measurements showed that the heavily worked structure contained a high proportion of HAGBs and that the material had a texture comprised mainly of brass {011}<211> and S {123}<634> components, as seen in Table 1. These

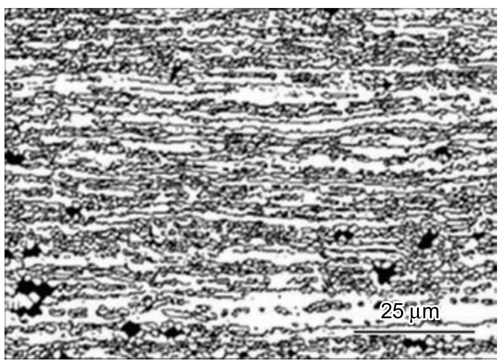

Fig. 1 EBSD map of the as-received microstructure from the RD-ND plane. Only high angle boundaries (>15°) are shown for clarity