TEXTURE DEVELOPMENT IN ALUMINIUM DURING HOT WORKING – EXPERIMENT AND SIMULATION

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Abstract

Samples of 99.5 % aluminium (AA 1050) were hot deformed by plane strain compression at 300° C and 400° C to strains of 0.5, 1.0, and 2.0 and subsequently quenched. Three different strain rates, $0.25\,s^{-1}$, $2.5\,s^{-1}$, and $25\,s^{-1}$, were used. The bulk textures were determined by neutron diffraction. The texture development was simulated using the Full and Relaxed Constraints models. The experimental and simulated results are described and compared. It is discussed how the correspondence between experiment and simulation could be improved taking into account microstructural observations.

Introduction

The development of deformation textures and microstructures depends on the processing parameters, and since the mechanical properties depend on the texture and microstructure, it is important to know the effects of the processing parameters. The present paper concentrates on the influence of *hot deformation* parameters on the *texture* development in aluminium.

The deformation textures of hot deformed aluminium alloys are the result of the simultaneous influence of deformation parameters such as deformation mode, temperature, strain, and strain rate. It is a rather complicated, if not impossible, task to separate the influence that the different deformation parameters have on the textural evolution for samples deformed in an industrial hot rolling mill. Plain strain compression (PSC) has proved to be a suitable tool for simulating industrial hot rolling, offering the possibility of varying strain, strain rate, and temperature independently [1].

In the present study the hot deformation textures in aluminium (AA 1050) have been investigated both experimentally and by simulation. The aim of the paper is to present a set of texture data that illustrates the influences of the deformation parameters separately and to evaluate to what extent the textural development may be simulated using standard models.

Experimental

The material under investigation was commercial purity aluminium (AA 1050) containing 0.32 % iron and 0.15 % silicon. It was prepared by industrial DC casting, homogenization at 600° C, slow cooling, break down rolling and annealing. The starting material contained 0.5 vol% 1.7 μm eutectic particles. Its average grain size (mean intercept length) was about 100 μm. The texture of the starting material was predominantly random with a small con-

tribution from the cube orientation. Approximately 8 vol%[1] had orientations near the cube position ($\{001\}\langle 100\rangle$). The three rolling components, namely, copper ($\{112\}\langle 111\rangle$), S ($\{123\}\langle 634\rangle$), and brass ($\{011\}\langle 211\rangle$) components occupied around 23 vol%.

The starting material was deformed by plane strain compression using the facilities at Pechiney Centre de Recherches de Voreppe. Samples were deformed at two temperatures $T = 300°$ C and 400° C, to strains (equivalent von Mises strain) $\varepsilon_{vM} = 0.5$, 1.0, and 2.0 applying three strain rates $\dot{\varepsilon} = 0.25\,s^{-1}$, $2.5\,s^{-1}$, and $25\,s^{-1}$.

For the texture measurements samples were taken from the mid third of the PSC samples. The textures were measured using the reactor DR3 at Risø National Laboratory [2], and the ODFs were calculated employing the series expansion method (e. g. [3]).

Simulation

The *Taylor-Bishop-Hill* theory [4,5] is the classical model for plastic deformation of polycrystals. It is based on the assumption that each crystallite undergoes a deformation identical to the macroscopic deformation of the polycrystal. Often it is referred as Full Constraints (FC) model. The classical *Taylor* model has been further developed into the so called Relaxed Constraints (RC) model [6,7] taking into account grain shape effects at higher strains. There are a number of other models (for an overview see for example [8]) but in the present study only the FC and RC models were applied.

For the simulations the commonly available packages *LApp: the Los Alamos Polycrystal Plasticity Simulation Code* [9] and *popLA: the Preferred Orientation Package from Los Alamos* [10,11] were used. In *popLA*, ODFs were calculated using the *Williams-Imhof-Matthies-Vinel* (WIMV) method [12].

The *LApp* code allows the incorporation of the initial experimental texture. Therefore the experimental ODF is converted to a set of discrete orientations (grains) in the *Euler* space. In the present work 1152 of such (weighted) orientations were used.

The deformation textures were simulated for strains $\varepsilon_{vM} = 0.2$, 0.5, 1.0, and 2.0, starting from the experimental initial texture using both the FC and RC models. For the purpose of comparison two more sets of calculations were performed with an initial random orientation distribution which will be referred as FC_{rnd} and RC_{rnd}.

Results and discussion

Experiment

A presentation of the experimental textures has been given in [13]. The results are briefly summarized here. The texture found after PSC is the typical copper type rolling texture. An example is shown in Fig. 1. The crystallite orientations with high frequencies lie along the β-fibre running from the copper over the S to the brass position. A certain amount of the Goss component is present as well.

All three rolling components strengthen at about the same rate when the strain is increased. In Fig. 2 the sum of the volume fractions of the three rolling components is plotted against the strain. The initially ($0 \leq \varepsilon_{vM} \lesssim 0.5$) high rate of texture development slows down at higher deformations ($\varepsilon_{vM} \gtrsim 1.0$). Increasing the deformation temperature from

[1]The volume fractions given here and below were obtained by integration over 15° around the corresponding peaks in the orientation distribution function (ODF) $f(g)$.

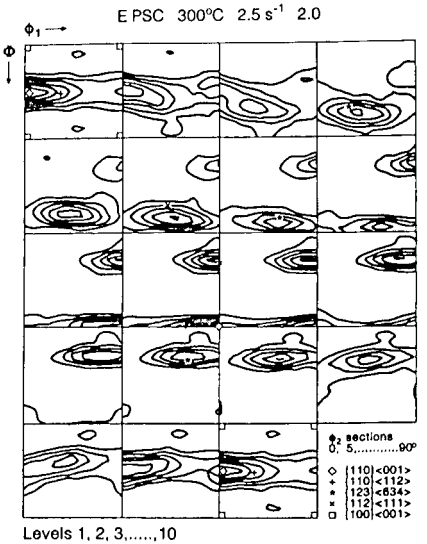

Figure 1: ODF of a PSC sample deformed at $T = 300°$ C with $\dot{\varepsilon} = 2.5\,\mathrm{s}^{-1}$ to a strain $\varepsilon_{vM} = 2.0$

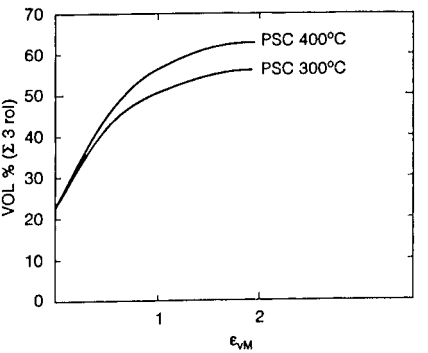

Figure 2: Development of the total volume fraction of the three rolling texture components (copper, S, brass) during PSC

$T = 300°$ C to $T = 400°$ C results in a stronger texture. The strain rate $\dot{\varepsilon}$, however, does not seem to have any systematic effects on the textural evolution[2]. The initially weak cube texture decreases with increasing strain. Within the limits of the experimental error, which typically amounts to a few per cent, neither temperature nor strain rate have a significant influence on the weakening as it seems.

At all strains the intensities along the β-fibre are almost constant, independent of φ_2 (see Fig. 3). The positions of the β-fibre in the *Euler* space are rather unaffected by strain and temperature (see Fig. 4). In the $\varphi_2 = 45°$ section the maximum is not situated at the exact copper position ($\varphi_1 = 90°$, $\Phi = 35°$) but at Φ approximately $25° - 27°$. Also the maximum in the $\varphi_2 = 65°$ section is shifted to a somewhat smaller value of Φ compared to the exact S position.

Simulation

In the simulated ODFs, the frequent orientations form a β-fibre (Fig. 5), and increasing strain gives stronger textures. At strains up to $\varepsilon_{vM} \sim 0.5$, $f(g)$ does not vary much with φ_2 along the β-fibre. At higher strains ($\varepsilon_{vM} \gtrsim 1.0$) however, one observes pronounced intensity changes along the β-fibre. This is illustrated by Fig. 3. A clear maximum is observed at $\varphi_2 = 50°$ (near the copper position) and local maxima are at the S and brass positions. It is

[2]Since no systematic influence of the strain rate $\dot{\varepsilon}$ on the textural evolution has been found for the description and discussion of experimental peak intensities and positions, here and below, the data obtained at $\dot{\varepsilon} = 2.5\,\mathrm{s}^{-1}$ was used.

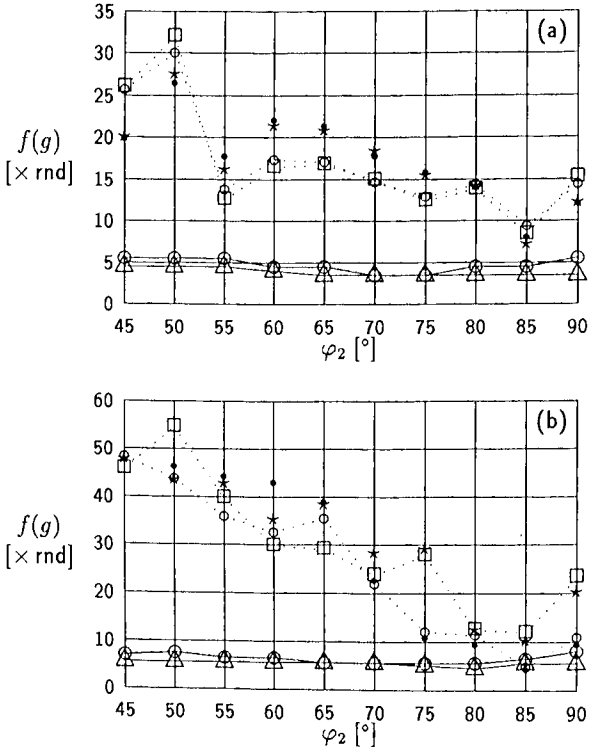

Figure 3: Values of the ODF $f(g)$ along the β-fibre in subspace I.
(a) Strain $\varepsilon_{vM} = 1.0$.
(b) Strain $\varepsilon_{vM} = 2.0$.
△ – experiment $T = 300°$ C, $\dot{\varepsilon} = 2.5\,\mathrm{s}^{-1}$
○ – experiment $T = 400°$ C, $\dot{\varepsilon} = 2.5\,\mathrm{s}^{-1}$
□ – FC
o – RC
⋆ – $\mathrm{FC_{rnd}}$
• – $\mathrm{RC_{rnd}}$

well known that the FC model does not give the exact copper position but a so called *Taylor* component at $\Phi \approx 27°$ approximately 8° from the copper peak ($\Phi = 35°$) (e. g. [14]). This can be seen in Figs. 4(e) and 4(g).

With increasing deformation the grains flatten more and more, and differences between the FC and RC models become visible. At $\varepsilon_{vM} = 1.0$ the FC and RC models differ already, but the deviations are still rather small and only for $\varepsilon_{vM} = 2.0$ would it be sensible to discuss the two models separately.

For the RC model at $\varepsilon_{vM} = 2.0$, the intensities vary along the β-fibre. The highest intensity is at the copper position, and there are local maxima at the S and brass positions. The values of $f(g)$ are of the same order of magnitude for the FC and RC models. The RC model develops somewhat higher intensities at the S and lower intensities at the brass positions than the FC model. Concerning the positions of the β-fibre, the RC model predicts a position closer to the exact copper position than the FC model. At a strain $\varepsilon_{vM} = 2.0$ the copper peak in the RC model has already moved $\Delta\Phi \approx 5°$ in the direction of the exact copper position (see Fig. 4(g)).

The differences between the calculations starting with a random ODF, $\mathrm{FC_{rnd}}$ and $\mathrm{RC_{rnd}}$, and the simulations using the initial experimental texture are mainly restricted to low strains. Differences in peak positions between FC and $\mathrm{FC_{rnd}}$ and between RC and $\mathrm{RC_{rnd}}$ are quite obvious for low strains ($\varepsilon_{vM} = 0.2$). With increasing deformation they disappear more and more, until the positions of the β-fibres for the respective "random" and "non random"

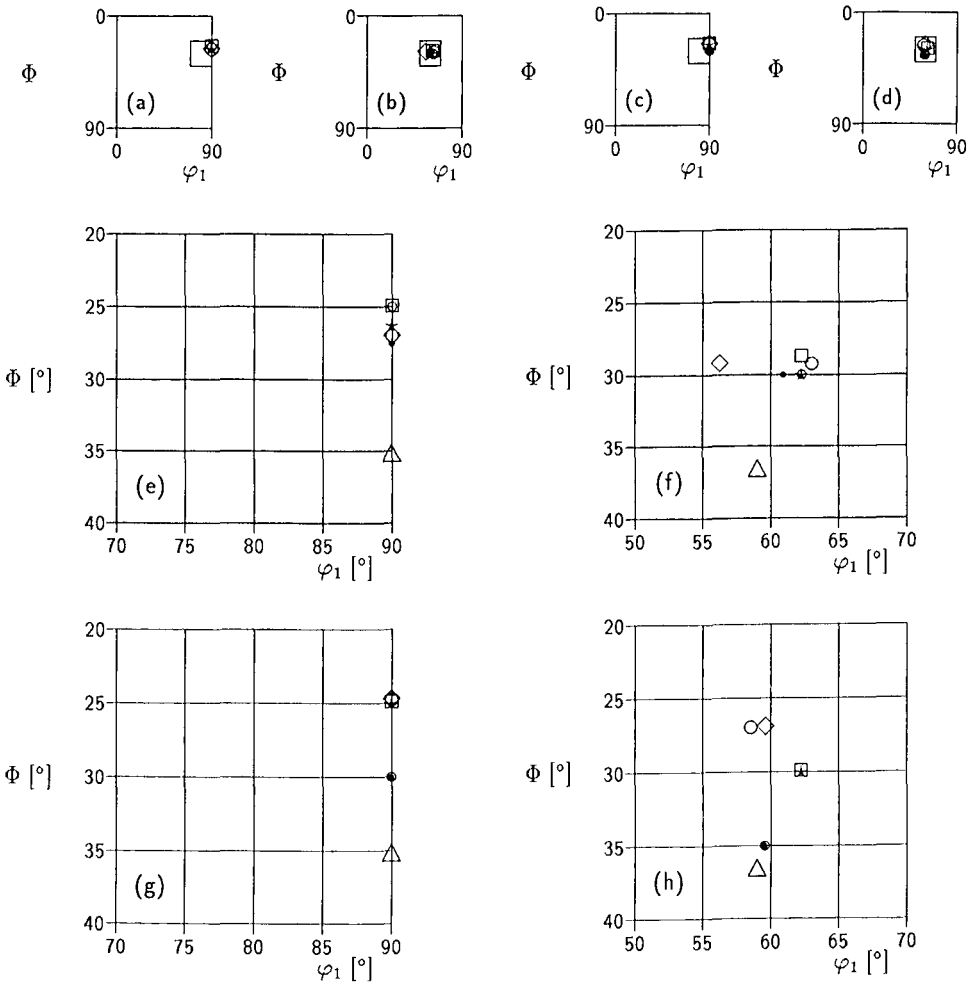

Figure 4: Positions of the β-fibre (subspace I) in the $\varphi_2 = 45°$ and 65° sections
(a) Strain $\varepsilon_{vM} = 1.0$; $\varphi_2 = 45°$. (b) Strain $\varepsilon_{vM} = 1.0$; $\varphi_2 = 65°$.
(c) Strain $\varepsilon_{vM} = 2.0$; $\varphi_2 = 45°$. (d) Strain $\varepsilon_{vM} = 2.0$; $\varphi_2 = 65°$.
(e) Strain $\varepsilon_{vM} = 1.0$; $\varphi_2 = 45°$ (detail). (f) Strain $\varepsilon_{vM} = 1.0$; $\varphi_2 = 65°$ (detail).
(g) Strain $\varepsilon_{vM} = 2.0$; $\varphi_2 = 45°$ (detail). (h) Strain $\varepsilon_{vM} = 2.0$; $\varphi_2 = 65°$ (detail).

$\diamond$ - experiment $T = 300°$ C, $\dot{\varepsilon} = 2.5\,\mathrm{s}^{-1}$ $\square$ - FC $\star$ - $\mathrm{FC_{rnd}}$
$\bigcirc$ - experiment $T = 400°$ C, $\dot{\varepsilon} = 2.5\,\mathrm{s}^{-1}$ $\circ$ - RC $\bullet$ - $\mathrm{RC_{rnd}}$
$\triangle$ - exact positions of the copper ($\varphi_2 = 45°$ section) and S ($\varphi_2 = 65°$ section) component

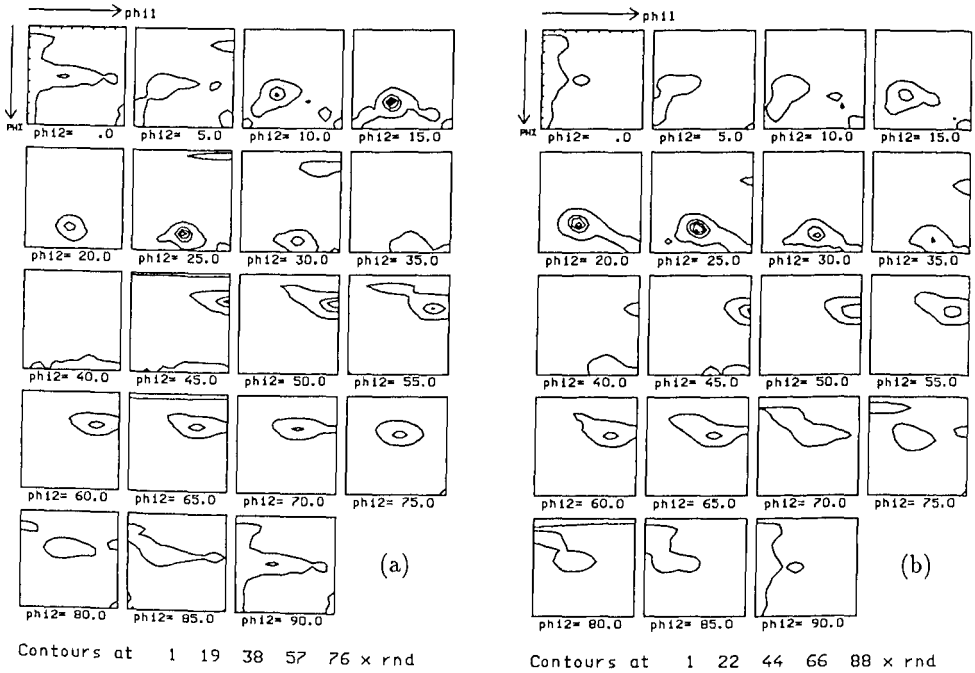

Figure 5: ODFs simulated for a strain $\varepsilon_{vM} = 2.0$ starting from the experimental texture (a) Full Constraints model. (b) Relaxed Constraints model.

cases almost coincide at $\varepsilon_{vM} = 2.0$ (see Fig. 4). However, effects of the initial texture on β-fibre intensities are visible even at strain $\varepsilon_{vM} = 2.0$ (compare Fig. 3). These differences are mainly restricted to the interval $45° \leq \varphi_2 \leq 65°$. The highest value of $f(g)$ for FC_{rnd} is situated at $\varphi_2 = 45°$ and there is a small peak at $\varphi_2 = 65°$ (S component) as well. The β-fibre for the RC_{rnd} calculation has a sigmoidal shape and no peak is found at $\varphi_2 = 65°$. In the simulations using an initial random ODF, higher orientation densities are, in general, obtained than in the calculations using the experimental (non random) starting texture.

A very important point when dealing with modelled textures is the necessity of smoothing the simulated ODFs [11]. In *popLA* a two-dimensional "*Gaussian* filter", i. e., a weighted averaging of neighbouring values with a specifiable decay distance is provided for this purpose. Non or weakly smoothed simulated ODFs are too "spiky" too be treatable. Very high degrees of smoothing distort the shape of the peaks more and more. All simulated ODFs have been smoothed with a smoothing parameter of 5°.

Comparison experiment – simulation

It is widely accepted that the FC model is a reasonably good first approximation for the texture development during deformation particularly at lower strains [8]. This applies also to the present investigation of hot deformation textures, where qualitatively a satisfactory correspondence between experimental and modelled ODFs is found (compare Figs. 1 and 5).

In general, there are no considerable differences of the β-fibre positions between experiment and simulation and it is difficult to detect any regularities in the deviations. For the present material, the FC model seems to predict the experimental results better than the RC model: At $\varepsilon_{vM} = 2.0$ the position of the experimental copper peak is better reproduced using the FC than the RC model (Fig. 4(g)). Also for the S component the FC model gives a better correspondence (see Fig. 4(h)).

The simulated textures are, however, much sharper and develop at a higher rate than the experimental ones as can be seen in Fig. 3. Already at a strain $\varepsilon_{vM} = 0.5$ the values along the β-fibre of the simulated ODFs are about twice the experimental ones. At higher strains the differences are even more pronounced. Figure 3 reveals also that the differences between experiment and model are much larger than the differences between the various types of simulations. Also the cube component decreases too quickly compared to the experimental observations.

It is generally observed that predicted textures are too sharp compared to the textures one obtains in experiment (see e. g. [8]). Quite obviously a stronger smoothing ("smearing") would result in textures with peak heights closer to the ones observed experimentally but at a price of more artificial peak shapes, especially, if very high smoothing parameters (e. g. more than $12.5° \cdots 15.0°$) are applied. Smoothing the data is, of course, a purely mathematical operation. There are, however, microstructural observations which could give a (stronger) smoothing a kind of physical "permission": In the *Taylor* model it is assumed that the grains deform homogeneously, whereas it is observed experimentally that the grains break up in volume elements - cell blocks delineated by dense dislocation walls (DDW) and microbands (MB) [15]. The DDW/MBs can be associated with large misorientations across them. This would be a *physical* "smearing" of the texture. To what extent the alloy under investigation shows "break up" under the different deformation conditions is not yet clear. For hot PSC deformed AA 3003 alloy it was found [16] that at any given strain the scatter of the texture in samples deformed at $T = 300°$ C was always larger than in those deformed at $T = 500°$ C. Microstructural studies by TEM have also shown that the deformation microstructure developing at 300° C is fairly similar to that after cold deformation with DDW/MBs, whereas at 500° C the microstructure is rather different with less DDW/MBs and thicker cell boundaries.

For the present AA 1050 material similar observations have been made: The full width half maximum (FWHM) values measured approximately perpendicular to the β-fibre axis at the copper and brass components were always larger at the lower temperature $T = 300°$ C, i. e., there is more scatter in the texture. TEM investigations were carried out for samples deformed 50 % by hot rolling (not by PSC). As in the case of the AA 3003 alloy, deformation at lower temperature (300° C) results in a microstructure similar to that obtained after cold deformation, whereas deformation at 400° C and 500° C results in coarser and more "clear" microstructures with less DDW/MBs. So, the observation of a larger texture spread at lower deformation temperatures can well be related to the type of microstructure developing at these temperatures, and a larger smoothing may be applied for simulation of the texture development at the lower temperature.

Conclusions

- The experimental texture after hot PSC is a typical copper type rolling texture. The three

rolling components strengthen and the initial cube component decreases with increasing strain. A higher deformation temperature leads to a stronger and less scattered texture. The strain rate seems not to influence the textural development in any systematic manner.

- The textures simulated with the standard FC and RC models give, in general, a qualitatively correct picture of the texture evolution observed in the experiments. However, both models are not capable to describe the textures quantitatively satisfactory. The calculated textures are much too strong and sharp and develop too fast compared to the experiment. The FC model gave a somewhat better prediction for the peak positions at high strains than the RC model.

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