MACROSCOPIC ANISOTROPY IN AA5019A SHEETS

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Abstract—The macroscopic anisotropy for typical texture components in aluminum alloys and AA5019A sheet samples (H48 and O temper conditions) were investigated. In order to simultaneously consider the effects of morphological texture and crystallographic texture on macroscopic anisotropy, predictions of plastic properties were carried out using a full-constraints Taylor model and a visco-plastic self-consistent (VPSC) polycrystal model. The yield stress and r-value (width-to-thickness plastic strain ratio in uniaxial tension) anisotropy predicted using the VPSC model were in good agreement with experimental data.

Keywords: Aluminum alloys; Mechanical properties (Texture; Constitutive equations; Plastic); Polycrystal modeling

1. INTRODUCTION

Macroscopic anisotropy of aluminum alloys has been widely studied due to its importance in process design. Microstructural parameters which affect the macroscopic anisotropy of aluminum alloy sheet or plate are the crystallographic texture [1, 2] and morphological texture, i.e. grain shape [3, 4], precipitate distribution [5, 6] and dislocation structure [4, 7]. As a result, it is necessary to investigate the individual and combined effects of these parameters on macroscopic anisotropy. To date, the prediction of macroscopic anisotropy for polycrystalline materials has been conducted using phenomenological [2, 8–21] and crystallographic approaches [22–34]. In the phenomenological approach, the plastic behavior of polycrystalline materials is assumed to be well described by analytical stress or strain rate potentials. The anisotropy coefficients of the potentials are determined using initial texture or experimental mechanical behavior data. These functions are easy to implement into FEM (finite element method) codes that simulate forming processes. However, sometimes the anisotropy coefficients are not sufficient to describe the macroscopic anisotropy of polycrystalline materials. In the crystallographic approach, the polycrystal is assumed to consist of many grains and plastic flow is assumed to occur only by crystallographic slip on active slip systems within each grain. Most of these studies are based on the full-constraints Taylor model [22]. The relaxed-constraint model [23–25] was introduced to overcome the shortcoming of the full-constraint Taylor model for the case of rolled materials exhibiting highly deformed grains. The model assumes lath or pancake-shaped grain morphology and does not satisfy equilibrium conditions between grains. Moreover, in real aluminum alloys during rolling deformation, development of lath or pancake-shaped grains is limited by local shear deformation or subgrain development. In order to understand the effect of crystallographic and morphological texture on the macroscopic properties more accurately, a more rigorous approach is required. Visco-plastic self-consistent (VPSC) polycrystal models [26–34] were recently developed to overcome this problem of the Taylor type model. Generally, these models assume a homogenization scheme in which the grain interactions with the matrix are taken into account. In the homogenization scheme, the material properties of a polycrystal aggregate vary from grain to grain and each grain is treated as a local inhomogeneity embedded in a homogeneous equivalent medium (HEM). By application of this homogenization scheme, Molinari et al. [26, 27] and Toth et al. [28, 29] predicted the texture evolution of f.c.c. and b.c.c. polycrystalline materials during rolling and torsion, respectively. Lebensohn et al. [30] predicted texture evolution during rolling and axisymmetric deformation of a zirconium alloy. They also calculated the plastic anisotropy of a rolled zirconium alloy sheet. However, no study has

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been reported to predict the macroscopic anisotropy in aluminum alloy sheet using a VPSC polycrystal model. In the present work, first of all, the effects of crystallographic and morphological texture on the macroscopic anisotropic properties (r-value and normalized yield stress) were investigated for typical crystallographic texture components in aluminum alloys using the full-constraints Taylor and VPSC polycrystal models. The isotropic crystallographic texture and typical components for rolling textures (Copper, Brass, S) and recrystallization textures (Cube and Goss) in aluminum alloys were considered. For the deformation textures, two different grain shapes (spherical and ellipsoidal) were analyzed. Moreover, the anisotropic properties of AA5019A sheets in the H48 and O temper conditions were measured and compared to predicted properties.

2. REPRESENTATION OF ORIENTATION SPREAD AROUND IDEAL COMPONENTS

Solidification, deformation, recrystallization and phase transformations all contribute to the development of crystallographic texture. After hot or cold rolling of aluminum alloys, the \{112\} < 111> (Copper), \{110\} < 11-2> (Brass) and \{123\} < 634> (Copper, Brass, S, Cube) are the dominant texture components. The relative intensity of these components depends on the rolling conditions and material composition. After annealing, depending on process conditions, the resulting textures can be nearly isotropic or mainly composed of the \{100\} < 001> (Cube) and \{110\} < 001> (Goss) recrystallization texture components. In the present work, the effect of five texture components (Copper, Brass, S, Cube and Goss) and an isotropic distribution of grain orientations on macroscopic mechanical properties is considered. Generally, a real polycrystal cannot be well-represented by a single or by a few ideal components, since a spread of orientations distributed around single ideal components is typically observed experimentally. Matthies et al. [35–37] have mathematically described an accurate orientation distribution function (ODF) around a single ideal component with the “Gaussian standard function” of the form

\[
f(S, \tilde{\omega}) = N(S) e^{S \cos \tilde{\omega}} \]

\[
N(S) = 1/[I_k(S) - I_t(S)]
\]

\[
I_k(x) = 1/\pi \int_0^\infty e^{\cos t} \cos kt \, dt
\]

\[
S = \ln 2/[2 \sin^2(b/4)] \quad \text{for} \quad b \leq 2\pi \quad (1)
\]

where \(f(S, \tilde{\omega})\) expresses the relative density of an orientation \(g\) rotated through an orientation distance of \(\tilde{\omega} = \tilde{\omega}(g^0, g)\) from the single ideal orientation \(g^0\). The terms \(b = 1.665\omega\) (\(\omega\) is the corresponding coefficient in Bunge’s notation [38]) and \(I_k(x)\) are the half-width (full width at half-maximum) of the bell-shaped curve and the modified Bessel function, respectively. In anisotropic materials, typical values of \(b\) are in the range 16.65–33.3° (or \(\omega\) from 10 to 20° [39]). An intermediate value of \(b = 25°\) (\(\omega = 15°\)) was selected from that range to describe the distribution of typical texture components of f.c.c. sheet metals in this study. Therefore, the textures evaluated in this work consisted of a 50% Gaussian distribution of orientations around an ideal component and a 50% isotropic texture. The ODF was used to generate a set of 1000 grain orientations for ideal textures and 2800 randomly selected orientations were used to represent an isotropic texture. For the spherical and ellipsoidal grains, 1–1–1 and 5–1–0.2 aspect ratios of the ellipsoidal axes (RD–TD–ND) were assumed, respectively.

3. EXPERIMENTAL

The experimental investigation was carried out on 0.22 mm thick AA5019A sheet samples in the H48 and O temper conditions. Mg and Mn are the main alloying elements in this alloy, with approximate contents of 5% and 0.3%, respectively. The final AA5019A–H48 samples were obtained after about 90% cold rolling reduction as a main deformation process. The grain morphology was investigated by optical microscopy. Figure 1(a) shows the microstructure of the RD/ND section of the AA5019A–H48 sheet as observed by optical microscopy. Although grains elongated in the rolling direction of the sheet can be observed, it is difficult to define the grain shape accurately in this heavily deformed sheet sample.

In the AA5019A–H48 sheets having high Mg content, shear localization can develop during severe rolling deformation [40]. Grains are probably broken up in many domains which can be assumed to be more equiaxed than the grains themselves [41]. Therefore, in this work, two types of grain shape were assumed for predicting the macroscopic properties of AA5019A–H48 sheet. If the grains deform homogeneously during cold rolling, their shape would be assumed to be ellipsoidal with a 10–1–0.1 aspect ratio (RD–TD–ND). However, considering shear localization, an approximate 5–1–0.2 aspect ratio (RD–TD–ND) was thought to be more reasonable. If equiaxed domains develop within each grain, the final domains are assumed to be spherical. In this case, the domains are assumed to play the role of embedded cells of uniform deformation in a self-consistent polycrystal model.

Pole figure measurements were carried out using a Seifert X-ray diffractometer for as-rolled AA5019A–H48 samples and for those exposed to
93°C (the thermal exposure of tensile test specimens, see later). From the \{111\} and \{200\} complete pole figures, the ODF was calculated using a harmonic series method. The \(\varphi_2\) sections of ODF for the cold rolled AA5019A–H48 specimen are shown in Fig. 2. The ODF was used to generate a set of grain orientations. This cold rolled material has a strong rolling texture and weak cube component. The orientation densities, \(f(g)\), on the \(\beta\)-fiber for the cold rolled AA5019A–H48 specimen and the AA5019A–H48 specimen exposed at elevated temperature are shown in Fig. 3. The characterization of the \(\beta\)-fiber shows that the sharp rolling texture components such as C (Copper) \(\{112\} < 11 \bar{1} 1 \rangle\), \(S \{123\} < 63 4 \rangle\), B (brass) \(\{110\} < 11 2 \rangle\) are dominant and that almost the same level of texture intensity is observed for both specimens. Therefore, the difference in texture for both materials (RT and exposure at 93°C during tensile testing) can be neglected.

Tensile specimens were cut from the sheet at angles of 0°, 15°, 30°, 45°, 60°, 75° and 90° to the rolling direction. Tension tests were conducted at room temperature (RT) and 93°C to measure the yield stresses and plastic strain ratios (\(r\)-value:width-to-thickness plastic strain ratio in uniaxial tension). Figure 4(a) shows engineering stress–engineering strain behavior measured in the 0°, 45° and 90° tension directions at RT and 93°C. The AA5019A–H48 sheets tested at RT exhibit serrated flow during tensile deformation. This phenomenon leads to difficulties in measuring the plastic strain ratio during tension tests at RT. Therefore, in order to overcome this problem in this work, tension tests were conducted at elevated temperature, 93°C, where serrated flow is suppressed. The AA5019A–O samples were obtained after about 90% cold rolling reduction, followed by annealing at 343°C for 4 h. Figure 1(b) shows the grain structure of the RD/ND section of AA5019A–O sheet observed by optical microscopy which reveals that this annealed material consists of almost spherical grains. The \(\varphi_2\) sections of ODF for AA5019A–O specimen are shown in Fig. 5. The ODF sections reveal that a strong cube and some retained rolling texture components are dominant.
Tension tests for these samples were conducted at RT to measure the yield stresses and plastic strain ratios (\(r\)-value). Figure 4(b) shows the uniaxial flow curves measured in the 0°, 45° and 90° tension directions at RT. Because the AA5019A–O specimens exhibited weak serrated flow during tensile deformation, the plastic strain ratios were measured without any difficulty.

4. ANISOTROPY PREDICTION

In order to predict the macroscopic properties of the polycrystalline aggregates, the Taylor (full constraints) polycrystal model and the VPSC polycrystal model were used. The VPSC polycrystal model was introduced to account for both texture and grain shape in the prediction of macroscopic prop-

Fig. 2. \(\phi_2 = 45^\circ\) sections of ODF for AA5019A–H48 sheet.
The shear rate on a given slip system \( g \), \( \dot{\gamma}_g \), is related to the resolved shear stress, \( \tau_s \), with a viscoplastic power law relation
\[
\frac{\dot{\gamma}_g}{\dot{\gamma}_{g0}} = \left( \frac{\tau_s}{\tau_{so}} \right)^m
\]
where \( m \) is the rate sensitivity parameter, \( \dot{\gamma}_{g0} \) and \( \tau_{so} \) are the reference shear rate and reference shear stress, respectively. The microscopic strain rate can be calculated as follows
\[
D_{ij} \hat{\gamma} \equiv \sum_s R_{ij} \left( \frac{R_{kl} S_{kl}}{\tau_{so}} \right)^{1/m}
\]
where \( D \) and \( S \) are the strain rate and the deviatoric stress tensors, respectively. The tensor \( R \) is associated with the slip system and can be expressed as follows
\[
R_{ij} = \frac{1}{2} (b_i n_j + b_j n_i)
\]
where \( b \) and \( n \) are the slip directions and the slip plane normal directions, respectively.

Using the concept of Homogeneous Equivalent Matrix (HEM) and 1-site approximation [26], the following interaction law between a grain and the matrix is derived
\[
S^g - \bar{S} = (\Gamma^g)^{-1} (D^g - \bar{D})
\]
where \( \bar{S} \) and \( \bar{D} \) are the deviatoric stress and the strain rate for the matrix. The effect of the grain shape can be taken into account by the interaction tensor \( \Gamma \) between single crystals \( g \) and \( g' \) whose components only depend on their shape. The tensor \( \Gamma \) can be obtained by integration of the symmetric part of the second derivative of the Green function.

For a plastically isotropic material, the macroscopic tangent modulus \( A \) is given as follows
\[
A_{ijkl} = \frac{2}{3} \frac{m I_{ijkl} S_{eq}}{D_{eq}}
\]
where the \( S_{eq} \) and \( D_{eq} \) are the von Mises equivalent stress and strain rate, respectively. \( I \) is the fourth rank identity matrix. The self-consistent model relies on the fact that \( A \) is calculated so that
\[
\{S\} = \bar{S} \quad \text{and} \quad \{D\} = \bar{D}
\]
where \( \{ \} \) means the arithmetic volume average over the polycrystal aggregate.

The \( r \)-value anisotropy and yield stress anisotropy were calculated using the Taylor (full constraints) and VPSC polycrystal models. For uniaxial tension test simulations, plastic strain rates were imposed...
5. RESULTS AND DISCUSSION

5.1. Ideal texture components

The $r$-value directionality calculated with the Taylor (full constraints) and VPSC (spherical and ellipsoidal VPSC) models is illustrated in Fig. 6 for each of the six calculated textures. While the Taylor (full constraints) and spherical-grain-shape VPSC

Fig. 5. $\varphi_2$ sections of ODF for AA5019A–O sheet.
models exhibit no directionality for the isotropic texture case, the ellipsoidal-grain-shape VPSC model exhibits some deviation from isotropic behavior. It should be noted that, for the isotropic texture [Fig. 6(a)], the morphological effect on the \( r \)-value directionality is reversed (\( r > 1 \) or \( r < 1 \)) at about 60° to the RD. For the Copper texture [Fig. 6(b)], the \( r \)-value directionality calculated with the Taylor (full constraints) model is stronger than that calculated with the VPSC model using either grain shape. The difference between \( r \)-value predictions from the two models is largest for tensile directions between 30° and 45° to the RD. For the Brass texture [Fig. 6(c)], the Taylor (full constraints) model shows a strong directionality for loading directions between 45° and 60° to the RD. This is consistent with a previous study on Al–Li sheet exhibiting a strong Brass texture component [34] where it was shown that the Taylor (full constraints) model overestimated the experimental.

Fig. 6. Comparison of \( r \)-value directionality predicted by the Taylor (full constraints) and the VPSC models for six texture components. (a) Isotropic; (b) Copper; (c) Brass; (d) S; (e) Cube; (f) Goss.
The $r$-value for directions between $45^\circ$ and $60^\circ$ to the RD. The $r$-value directionality for the S texture [Fig. 6(d)] is similar to that of the Brass texture. Among the deformation textures, the morphological effect of the grain shape on the $r$-value directionality is the weakest for the Copper texture.

Figures 6(e) and (f) show the results of $r$-value directionality for the Cube and Goss crystallographic texture components. Because Cube and Goss components are associated to recrystallization textures with roughly equiaxed grains, only the spherical grain shape was considered to predict the macroscopic properties. For these specific cases, $r$-value directionality calculated with the Taylor (full constraints) model is stronger than that calculated with the spherical-grain-shape VPSC model. The difference between the two models for the Cube texture is the most significant for tensile directions between $30^\circ$ and $60^\circ$ to the RD. For the Goss texture [Fig. 6(f)], the spherical-grain-shape VPSC model predicts $r$-values that are significantly lower than that those predicted with the Taylor (full con-

Fig. 7. Comparison of normalized yield stress directionality predicted by the Taylor (full constraints) and the VPSC models for six texture components. (a) Isotropic; (b) Copper; (c) Brass; (d) S; (e) Cube; (f) Goss.
straints) model for directions between 60° and 90° to the RD. However, both models predict almost the same \( r \)-values for loading directions between 0° and 60° to the RD.

The normalized yield stress predicted by the Taylor (full constraints) and VPSC (spherical and ellipsoidal-grain-shape VPSC) models are shown in Fig. 7 for each texture component. For the isotropic texture [Fig. 7(a)], the morphological effect of the ellipsoidal grain shape leads to continuously increasing yield stress from RD to TD. In the case of the deformation textures, the yield stress anisotropy predicted with the Taylor (full constraints) model is weaker than that predicted with the VPSC model. However, it should be noted that the morphological effect (ellipsoidal-grain-shape) leads to higher strength anisotropy for all loading directions.

Comparing the Copper and Brass texture results [Figs 7(b) and (c)], it is observed that the morphological effect on the yield stress is reversed, i.e., the ellipsoidal grain shape leads to higher yield stresses for Brass and lower yield stresses for Copper textures. This explains why, for a polycrystal material having Copper and Brass texture components simultaneously, the morphological effect on the yield stress anisotropy is negligible. Among the three deformation textures studied here, the S texture has the lowest morphological effect on the yield stress directionality. In contrast with the results obtained with deformation textures, it should be noted that the yield stress anisotropy predicted from the VPSC model for the recrystallization textures [Figs 7(e) and (f)] is weaker than that predicted with the Taylor (full constraints) model.
5.2. AA5019A sheet

Figure 8(a) shows the r-values for AA5019A–H48 as calculated with the polycrystalline models in different directions as well as those determined experimentally. The predicted r-values using the Taylor (full constraints) model overestimate the experimental data for most of the tensile directions. The VPSC calculations assuming ellipsoidal grain shapes exhibit good agreement with experimental values within the range of experimental errors. Figure 8(b) shows the normalized yield stresses calculated with the polycrystal models as well as those determined experimentally. The Taylor model underestimates the experimental data for loading directions between 0° and 55° to the rolling direction (RD), and overestimates the experimental data for the directions greater than 55° to the RD. The VPSC model underestimates the experimental data for all loading directions. The difference between the VPSC model and the experimental results could be due to texture gradient effect through the thickness of the sheet or due to microstructural effects such as dislocation structure. An additional model considering the deformed initial dislocation structure could be implemented in this calculation to make the initial critical resolved shear stress dependent on the orientation of the slip systems with respect to the sheet sample reference frame.

Figure 9(a) shows the r-values for AA5019A–O sheet as calculated with the Taylor (full constraints) and VPSC polycrystalline models in different directions as well as those determined experimentally. In the present work, a spherical grain shape was assumed for predicting the macroscopic properties of the AA5019A–O sheet using the VPSC model. The predicted r-values using the Taylor (full constraints) model underestimate the experimental data for most of the tensile test directions, whereas the predicted r-values using the VPSC model are in good agreement with the experimental trend. The VPSC calculations assuming spherical grain shapes exhibit good agreement with experiments within the range of experimental errors. Figure 9(b) shows the normalized yield stresses calculated with the polycrystal models as well as those determined experimentally. It should be noted that the yield stress anisotropy calculated from the Taylor (full constraints) model predicts behavior which is opposite to that of the experimental data for all loading directions. The predicted normalized yield stresses using the VPSC model underestimate the experimental data for all loading directions, but these predictions exhibit almost the same trend in yield stress anisotropy as the experiments. According to the above results, the VPSC model which assumes spherical grain shape can predict the macroscopic anisotropy better than the Taylor (full constraints) model.

6. CONCLUSIONS

1. For the deformation textures (Copper, Brass, S), the yield stress and r-value anisotropies calculated with the Taylor (full constraints) model are, respectively, weaker and stronger than those calculated with the VPSC model. The ellipsoidal grain shape is predicted to have higher r-value anisotropy and yield stress anisotropy, simultaneously, than the spherical grain shape for deformation and isotropic textures. For the recrystallization textures (Cube, Goss), the r-value and yield stress anisotropy calculated with the Taylor (full constraints) model are stronger than those calculated with the VPSC model.

2. For an AA5019A–H48 sheet sample, the Taylor model overestimated experimental r-values, particularly for tension loading directions near 45° to the RD. The VPSC model assuming ellipsoidal grain shape resulted in excellent predictions in all directions. For the normalized yield stress of AA5019A–H48, both the Taylor (full constraints) and VPSC models showed deviations compared with experimental results. Predicted grain shape effects using the VPSC model for yield stress directionality were negligible. In AA5019A–O, the Taylor (full constraints) model not only underestimated r-values particularly for loading at 45° to the RD, but it also predicted an opposite anisotropy in yield stress. The results of the VPSC model were in good agreement with experimental results. In this model, the texture was used as an input and the grains were assumed to have a spherical shape for the O temper sheet.

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