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EFFECT OF CRYSTALLOGRAPHIC AND MORPHOLOGIC TEXTURES ON THE ANISOTROPY OF MECHANICAL PROPERTIES OF Al-Li ALLOYS

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ABSTRACT
A new formulation of the self consistent model has been applied to predict the plastic behaviour of Al-Li Alloys. This model enables to take into account simultaneously the crystallographic texture, the morphology of grains, the internal stresses evolution and the anisotropy of intracrystalline hardening. Microscopic observations and analyses of crystallographic textures are used to constitute the data for the self consistent scheme. As results of such calculations, we present the variations of plastic properties versus the angle between the tensile axis and rolling direction. Initial Yield point, macroscopic hardening modulus as well as transversal plastic flow are examined. The direct comparison with experimental results is given. A satisfactory agreement with numerical results is observed.

I. INTRODUCTION
The recent development of Al-Li Alloys corresponds to the increasing demand of the low density and high strength materials for aircraft applications. These alloys are usually furnished as plates or sheets obtained by rolling and exhibit various anisotropies of mechanical properties such as:
- anisotropy of the yield point
- anisotropy of the macroscopic hardening modulus
- anisotropy of the transversal plastic flow.

The macroscopic anisotropy is closely related to the anisotropic behavior of the single crystal and the microstructure of the grains:
- the anisotropy of the elastoplastic behavior of the single crystal results from the crystallographic nature of plastic glide, interactions between the glide systems (reflecting the dislocation - dislocation and dislocation - precipitation interactions) and eventually anisotropy of elastic constants.
- the mechanical anisotropy associated with the granular aspect of the metal is caused by the crystallographic texture, morphologic texture (shape of grains) and internal stresses of the second order engendred by the intragranular incompatibilities of the plastic strain.

In order to predict the anisotropic behavior of Al-Li alloys, all the micro-physical phenomena mentioned above should be taken into considerations. The modelling of the anisotropy of Al-Li alloys was developed by H. Peters, K. Welpmann and T.H. Sanders [1]. Their model based on the Schmid's law governing the activation of glide systems approached the polycrystal by a set of independent grains. This approach neglects different sources of the plastic anisotropy of the polycrystal excepting for crystallographic texture.

In this paper, we use the self consistent method to predict the behavior of Al-Li alloys. The obtained results are directly compared with the experimental measurements performed on sheets in 291 alloys. The sheets, of 3.6 mm thickness, have all been tested in the T8x51 temper. First, the fundamental hypotheses of the method are rewiewed and next,
the way to introduce the important micro-physical parameters into the model is discussed.

II. MODELLING OF THE SELF CONSISTENT METHOD

1) Single crystal behavior

It is supposed that the crystallographic glide is the main micromechanism of plastic strain of the F.C.C.-crystal. Let \( \dot{\gamma}_g \) be a plastic resolved shear strain rate on a g-th slip system characterized by \( \vec{n}_g \) unit vector normal to the slip plane and \( \vec{m}_g \) unit vector of slip direction. The plastic strain rate \( \dot{\varepsilon}_p \) resulting from simultaneous slips on active systems is given by

\[
\dot{\varepsilon}_p = \sum_g R_{ij}^g \dot{\gamma}_g
\]

with

\[
R_{ij}^g = \frac{1}{2} (m_i^g \vec{n}_g + m_j^g \vec{n}_g)
\]

A system \( g \) becomes potentially active if the following relation, called Schmid law, is verified:

\[
\tau_c^g = R_{ij}^g \sigma_{ij}
\]

where the threshold stress \( \tau_c^g \), named critical resolved shear stress is supposed to evaluate during plastic strainings. If a linear strain hardening is described by a hardening matrix \( H^h \), the critical shear stress rate on a system labeled \( h \) can be expressed by

\[
\dot{\tau}_c^h = \sum_g H_{kg}^h \dot{\gamma}_g
\]

On the other hand, the resolved shear stress rate on a system \( g \) is obtained knowing the local stress rate \( \sigma_{ij} \) by

\[
\dot{\gamma}_g = R_{ij}^g \dot{\sigma}_{ij}
\]

A system \( g \) is active if simultaneously (2) is verified and \( \dot{\sigma}_{ij} \) is such that \( \dot{\gamma}_g > 0 \).

Knowing all active systems and the local stress rate, the plastic strain rate may be determined from the above equations

\[
\dot{\varepsilon}_{ij} = \sum_{g,h} R_{ij}^g K_{gh}^h R_{kl}^h \dot{\sigma}_{kl}
\]

where

\[
K_{gh}^h = (H^h)^{-1}
\]

If the elastic constitutive law is expressed by the following relation

\[
\dot{\varepsilon}_{ij} = s_{ijkl} \dot{\sigma}_{kl}
\]

the formal constitutive equation for the single crystal may be written down in the form

\[
\dot{\sigma}_{ij} = (s_{ijkl} + \sum_{g,h} R_{ij}^g K_{gh}^h R_{kl}^h) \dot{\varepsilon}_{kl}
\]

or, inverting (7), as

\[
\dot{\varepsilon}_{ij} = l_{ijkl} \dot{\varepsilon}_{kl}
\]
2) Behavior of the polycrystal

On the macroscopic level, the constitutive equation of the polycrystal is defined by the relation

\[ \dot{\varepsilon}_{ij} = L_{ijkl} \dot{E}_{kl} \]  

(9)

Where \( \dot{\varepsilon} \) is the overall strain rate tensor (\( \dot{\varepsilon} = < \varepsilon > \)), \( \dot{\varepsilon} \) is the overall stress rate tensor (\( \dot{\varepsilon} = < \sigma > \)), and \( L_{ijkl} \) is the tensor of the overall instantaneous moduli which reflects both the anisotropy of the single crystal and those due to the microstructure of polycrystal (crystallographic and morphologic texture).

It has been shown [2], that it exists an integral relation linking \( \dot{E}_{kl} \) and \( \varepsilon_{kl} \). This integral equation may be solved using a self consistency hypothesis. The solution of the integral equation can be written as follows:

\[ \varepsilon_{ij} = A_{ijkl} \dot{E}_{kl} \]  

(10)

where \( A_{ijkl} \) is the concentration tensor for the strain rate tensor. The components of \( A \) depend on:

- the mechanical anisotropy of the polycrystal \( L^{\text{eff}} \)
- the behavior of grain \( l(\Omega) \) of a given crystallographic orientation defined by \( \Omega \)
- the shape of grains.

The various methods of determination of the tensor \( A \) may be found in [2]. Here we mention only that the Lin-Taylor [3] model is obtained putting \( A = 1 \) which expresses the homogeneity of total local strain rate. If additionally the elastic strain is neglected, the Taylor model [4] is obtained.

In order to determine the tensor \( L^{\text{eff}} \), let's substitute (10) into (8)

\[ \dot{\varepsilon}_{ij} = l_{ijkl}(\Omega) A_{klmn}(\Omega) \dot{E}_{mn} \]  

(11)

Now, using the averaging operation over the polycrystal volume, the above equation leads to

\[ L_{ijkl}^{\text{eff}} = \frac{1}{V} \int_V l_{ijmn}(\Omega) A_{mnkl}(\Omega) \, dV \]  

(12)

Introducing the orientation distribution function \( f(\Omega) \) such that

\[ \frac{dV(\Omega)}{V} = f(\Omega) \, d\Omega \]

equation (12), can be rewritten as

\[ L_{ijkl}^{\text{eff}} = \int_{\Omega} l_{ijkl}(\Omega) A_{klmn}(\Omega) f(\Omega) \, d\Omega \]  

(13)

This relation takes into account the influences of the crystallographic texture of the polycrystal on the overall behavior of the aggregate. All the sources of the single crystal anisotropy are taken into account by this equation. The elastic constant anisotropy is reflected by the tensor \( s_{ijkl} \), the plastic anisotropy due to the crystallographic nature of deformation modes is rendered by \( R \) tensors and finally the anisotropy of the strain hardening is modelled by the hardening matrix \( H \). The shape of grains (morphological textures) is taken into account by evaluating the \( A \) tensor for ellipsoidal inclusions.

III. APPLICATIONS OF THE MODEL TO Al-Li ALLOYS

In order to perform the numerical simulations, the elastic and plastic properties of the single crystal should be specified. The elastic behavior of the single crystal and consequently those of the polycrystal are supposed to be isotropic and defined by Poisson's
ration \( v = 0.3 \) and Lamé's modulus \( \mu = 3000 \text{ daN/mm}^2 \). The plastic slip on the twelve easy glide systems of \(<111> (110)\) type is governed by the Schmid's law. The initial critical shear stress is supposed to be the same for all glide systems of each grain and equals \( \tau_o = 12.5 \text{ daN/mm}^2 \). This value is deduced from the experimental data. The hardening rule of the single crystal is described by a constant matrix \( H \) composed of two types of terms. The first term modelling the weak interactions between the glide systems is equal to \( \mu/250 \). The second term, being 3 times greater reflects the strong interactions between dislocations. This matrix describes well the dislocations - dislocations interactions. The dislocations - precipitate interactions which can considerably modify the single crystal behavior are only roughly approached for reason of the lack of numerical informations.

Finally, the shape of grains was estimated using the optical metallography. The results of these observations lead to the following choice of semi axes of grains:

\[
\begin{align*}
a &= 15 \mu\text{m} \\
b &= 5 \mu\text{m} \\
c &= 1 \mu\text{m}
\end{align*}
\]

The same inclusion shape has been chosen for all grains and such that; \( a \)-axis indicates the rolling direction \( b \)-axis transversal direction and \( c \)-axis the normal to the sheet.

The crystallographic texture was measured by X-Ray diffraction and analysed using the vectorial method [5]. The results of these measures are presented in details in another communication at this conference [6]. The numerical simulations have been performed using a certain number of ideal components of the experimental texture. The corresponding weights have been estimated from the orientation distribution function using various standard deviations of integration. The results are presented in Table I.

<table>
<thead>
<tr>
<th>Standard deviation</th>
<th>Goss (%)</th>
<th>S (%)</th>
<th>Cube (%)</th>
<th>Copper (%)</th>
<th>Brass (%)</th>
</tr>
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<tbody>
<tr>
<td>5°</td>
<td>1,04</td>
<td>0</td>
<td>0</td>
<td>0</td>
<td>1,88</td>
</tr>
<tr>
<td>10°</td>
<td>1,04</td>
<td>0</td>
<td>0</td>
<td>0,22</td>
<td>6,44</td>
</tr>
<tr>
<td>15°</td>
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<td>0,9</td>
<td>1,28</td>
<td>0,92</td>
<td>9,86</td>
</tr>
<tr>
<td>20°</td>
<td>6,08</td>
<td>6,98</td>
<td>3,64</td>
<td>2,61</td>
<td>15,29</td>
</tr>
</tbody>
</table>

TABLE I : Volumic fractions of five ideal texture components.

The experimental tensile tests have been performed at the room temperature under constant strain rate conditions \( \dot{\varepsilon} = 2 \times 10^{-4} \text{ s}^{-1} \). Samples were taken from the rolled sheets for various \( \alpha \) angle measured with respect to the rolling direction. The corresponding tension curves are drawn in fig. 1. An important yield point anisotropy is observed and presented on fig. 2. Fig. 3 shows the anisotropy of a hardening moduli defined as \( \Delta \sigma /\Delta \varepsilon \) and taken for plastic strain internal given by \( \varepsilon^p = 1\% \) and \( \varepsilon^p = 2\% \).

The corresponding numerical results are plotted on figs. 4, 5, and 6. Fig. 4 shows the obtained tension curves (evolution of \( \Sigma_{11} \) - stress versus \( \varepsilon^p_{11} \) - plastic strain). Fig. 5 presents the influence of the \( \alpha \) angle and plastic offset definition on the yield point. Finally, fig. 6 summarizes the anisotropy of the yield point of three main texture components for two plastic offsets namely \( \varepsilon^p = 0.025\% \) and \( \varepsilon^p = 0.2\% \). The comparison of the theoretical and experimental results indicates a good qualitative agreement concerning the evolution of the yield point anisotropy. Nevertheless, certain differences concerning the hardening evolution and its anisotropy may be remarked.
Fig 1. Experimental tension curves for various values of \( \alpha \) angle between the rolling and tension directions.

Fig 2. Experimental anisotropy of yield points versus \( \alpha \) angles for various plastic offset definitions.

Fig 3. The anisotropy of the hardening modulus in function of the tension direction.
Fig 4. Numerical simulations of tension curves for various $\alpha$-angles.

Fig 5. Theoretical evolution of the anisotropy of yield point versus $\alpha$-angle for four plastic offsets.

Fig 6. The influence of the tension direction on yield point of three ideal texture components
   a) plastic offset $E^P = 0.025\%$
   b) plastic offset $E^P = 0.2\%$
These differences are mainly due to a poor description of crystallographic textures by intermediary of a few ideal components. Moreover, the anisotropy of the single crystal hardening strongly depends on dislocation – precipitate interactions. These interactions have been only roughly modelled in the calculations. A more complete study of these problems is undertaken.

REFERENCES


[2] P. Lipinski, M. Berveiller
Elasto-plasticity of micro-inhomogeneous metals at large strains. Submitted for publication


Quantitative determination of the texture of Al-Li Alloys (This conference)