Anisotropic plasticity in oriented semi-crystalline polymer systems

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Abstract

In this thesis a viscoplastic model is proposed to characterize the macroscopic off-axis yield behavior of oriented semi-crystalline polymers. The viscoplastic model is defined in a general form, such that different flow rules and equivalent stress definitions can be inserted. An equivalent yield stress, based on Hill's anisotropic yield criterion, is used with the Eyring theory in combination with a viscoplastic flow rule. This model is capable of quantitatively predicting the rate and angle dependence of the yield stress as well as time to failure in various off-axis loading.

The micro structural origin has been investigated and it is shown that the contribution of the crystalline and amorphous phase to the yield stress in the 20° and 90° loading direction are additive in stress and the 0° loading direction in strain rate. Moreover, it is shown that the lamellar thickness does not influence the deformation kinetics.
Chapter 1

Introduction

The widespread use of polymers in engineering applications has led to an increased demand for understanding of the mechanical behavior of these materials, especially regarding the yield or failure behavior. The ability to model and predict the behavior of these materials is desired.

Efforts of modeling begin with classical works of von Mises\(^1\) and Tresca\(^2\) for isotropic polycrystalline materials. Both of these theories consider the yield stress of the material to be independent of strain rate, temperature and pressure. For polymers, however, the effects of temperature and strain rate on the yield behavior of polymeric materials are significant and therefore these must be considered in an appropriate criterion for plastic yielding.\(^3-6\) The yield stress of an isotropic polymer has been found to increase with increasing strain rate and decreasing temperature. In general, the plots of yield stress versus the logarithm of strain rate show either a straight line or a significant slope change indicating simple or complex thermally activated Eyring flow processes respectively.\(^7,8\) For isotropic polymers, 3D models have been developed which capture the pressure, strain rate and temperature dependence of the yield stress.\(^9-11\)

Polymer products are often manufactured by common processing techniques such as extrusion and injection molding. Within these processes, in case of a semi-crystalline polymer, the molten material is subjected to shear and elongational flow, prior to crystallization which typically induces a degree of orientation. Therefore injection moulded and extruded samples of various semi-crystalline polymers, such as polyethylene and polypropylene, show a different morphology than isotropic polymers. For semi-crystalline polymers, orientation of the crystals towards the flow direction has been observed.\(^12\) Semi-crystalline polymers also possess remarkable drawability, which makes it possible to highly orient them in the solid state. During solid state processing, macromolecular chains are often preferentially oriented along a specific direction, e.g., the drawing direction, so that the final products possess the desired properties in a particular direction.\(^13\) As a result, the yielding behavior of these polymers becomes anisotropic and therefore yield criteria should be adapted to account for this.

The most common yield criterion to deal with anisotropy was developed by Hill.\(^14\) In general, the yield stress decreases with increasing loading angles with respect to the initial drawing direction. Most studies have not investigated the pronounced influence of varying strain rates and temperatures, with the exception of the work of Kartunnen and Lesser\(^15-20\). They extended the Hoffman criterion, which is based on Hill's criterion, to account for strain rate and temperature. The model
predicts the shift in yield stress, from a reference state, of an aliphatic polyketone terpolymer as a function of strain rate and temperature.

In the present work uniaxial tensile and long-term failure tests are performed on solid state drawn polypropylene tape to investigate the effects of strain rate and temperature. A viscoplastic model is proposed in a general form, such that different flow rules and equivalent stress definitions can be inserted, to predict off-axis behavior of oriented polypropylene tapes. An equivalent yield stress, based on Hill's anisotropic yield criterion, is used with the Eyring theory in combination with a viscoplastic flow rule. The model predictions are then compared with experimental yield stress data.
Chapter 2

Experimental

Materials

The semi-crystalline polymers used in this study are high density polyethylene (HDPE) and polypropylene (PP). For PE (DSM, Stamylan HDPE) two grades are selected, which are the same as in the work of Schrauwen. The characteristics of these grades are given in table 2.1.

<table>
<thead>
<tr>
<th>Material</th>
<th>$M_w$ [g mol$^{-1}$]</th>
<th>$M_n$ [g mol$^{-1}$]</th>
</tr>
</thead>
<tbody>
<tr>
<td>PE 8621</td>
<td>70000</td>
<td>11000</td>
</tr>
<tr>
<td>PE 9089S</td>
<td>210000</td>
<td>7000</td>
</tr>
</tbody>
</table>

Table 2.1: Characteristics of polymer grades used in this study.

The PE was compression moulded with a two step protocol resulting in a homogenous sample with a low surface roughness. Dogbone shaped samples were cut out of the compression moulded sheets. The samples were subsequently subjected to different crystallization procedures in order to obtain different levels of crystallinity and lamellar thickness. The procedures applied consisted of either placing the molten samples in a cold press of 15°C (quenching, Q15), leaving the molten samples in the press turning off the heat source (slow cooling, SC), allowing to cool to room temperature or annealing in a hot oven at 120°C for 8 hours (isothermal crystallization, IC120), which was only applied to the 'quenched' samples.

An isotactic PP (Dow iPP homopolymer resin H507-03Z) tape used in this paper was monoextruded using a Collin E20-T, 20 mm, single screw extruder fitted with a 100 mm die. The die lip was adjusted to a gap of 1 mm. The temperature profile of the extruder from feed zone to die was: 170, 180, 190, 200°C. The extrudate was quenched on Collin CR-72 rolls at a temperature of 15°C and then collected on a spool. Solid state drawing was performed off-line on Collin MDO-AT and BT roll stands using a short gap stretching regime. The drawing temperature was 120°C, with the draw ratio being defined as the speed ratio of the inlet vs. the outlet goddet. A schematic of the set-up is shown in figure 2.1. Different iPP tapes are produced with increasing draw ratios of $\lambda = 1$, 4 and 6. The tape with a draw ratio of $\lambda = 1$ is found to be mechanically isotropic.
Figure 2.1: Schematic of the extrusion and subsequent drawing set-up. Extrusion settings for zones 1 to 5 are 170, 180, 190, 200, 200°C, resp. and drawing line settings of zones 6 and 7 are 120, 25°C, resp. Zone 0 is water cooled.

Dogbone-shaped samples, with dimensions of the smallest section of 20 x 2 mm, are cut directly from the tapes at different angles, 0°, 5°, 10°, 15°, 20°, 30°, 45°, 75° and 90°, with respect to the drawing direction (DD), see figure 2.2.

Figure 2.2: Orientation of tensile testing samples with respect to the drawing direction (DD) in a polypropylene tape.

SAXS / WAXS

The degree of crystallinity and the lamellar thickness are measured by wide- and small angle X-ray scattering experiments, respectively. Samples are used to measure simultaneously 2D SAXS and 1D WAXS patterns at the Dutch-Belgian beamline (DUBBLE, CRG BM26) at the European Synchrotron Radiation Facility (ESRF)
in Grenoble, France. The size of the beam was 1 x 0.3 mm$^2$, the X-ray wavelength was 1.24 Å and the sample-to-detector distance for SAXS experiments was 5m. The 1D WAXS detector was calibrated using the known peak positions of HDPE for the given wavelength using Bragg’s law:

$$n\lambda = 2d\sin\theta$$

(2.1)

where $n$ is an integer determined by the order given, $\lambda$ is the wavelength, $d$ is the spacing between the planes in the atomic lattice and $\theta$ is the angle between the incident ray and the scattering planes. Figure 2.3(a) shows an example of a WAXS profile of an amorphous and semi-crystalline sample. Background scattering is subtracted from the profiles in order to determine the degree of crystallinity of the semi-crystalline sample using:

$$\chi = \frac{C - A}{C}$$

(2.2)

where $C$ is the area of the crystalline profile and $A$ is the area of the amorphous profile.

The amorphous halo is approximated with a 5th order polynomial. The crystalline long period, $D$, which is related to the distance between lamellae, is determined from the SAXS measurements. The patterns are integrated in radial direction, resulting in 1D SAXS profiles, see for example figure 2.3(b). Intensity is plotted against the scattering vector $q$, which is defined as:

$$q = \frac{2\pi}{D} = \frac{4\pi}{\lambda} \sin\theta$$

(2.3)

The scattering angle is calculated from the position of the scattered intensity on the detector and the sample-to-detector distance, which was calibrated accurately using the scattering pattern of dry rat tail collagen.

The peak positions of the SAXS profiles are taken to calculate the average lamellar thickness from:

$$L = \chi \cdot D$$

(2.4)

where $\chi$ is the degree of crystallinity obtained from the WAXS profiles.
Figure 2.3: Example of WAXS (a), in which the amorphous (A) and crystalline (C) part are indicated, and SAXS profiles (b) of PE samples with different crystallization procedures applied.

**Mechanical testing**

Uniaxial tensile measurements are performed on a Zwick Z010 universal tensile tester, equipped with a 2.5 kN cell, at ambient conditions (22°C). More tensile tests are carried out on a servo-hydraulic MTS Elastomer Testing System 831 equipped with a 2.5 kN force cell and an environmental chamber. These tests are conducted at two elevated temperatures, 40°C and 60°C.

Tensile experiments are performed at a constant linear strain rate of $10^{-4}$, $10^{-3}$ or $10^{-2}$ s$^{-1}$. The engineering stresses are calculated using the average of the cross-sectional areas as measured at three locations in the gauge length.

Each experiment is repeated three times and the mean values of the yield stresses are reported. For small loading angles, with respect to drawing direction, no yield is observed and hence the ultimate stress is denoted as the yield stress. For loading angles from 15° to 30°, no visible softening, i.e. clear yielding, is detected. Subsequently, via the tangent method yield stresses are calculated. For loading angles larger than 30°, a stress-strain curve is measured with clear strain softening and subsequent strain hardening. For these samples, the maximum in stress before softening is taken as the yield stress.

Long-term failure experiments are performed on the Zwick Z010 device under load control, applying a constant load in uniaxial tension. Endpoints of these experiments are the time-to-failure. In all long-term failure experiments, the time needed to apply the constant load was negligible in comparison to the time-to-failure obtained.
Chapter 3

Anisotropic plastic flow

3.1 Phenomenology

The macroscopic behavior of solid state drawn polypropylene tapes is investigated to detect the influence at room temperature of draw ratio, off-axis loading angle and strain rate. First, the observed macroscopic behavior is discussed. Figure 3.1 shows the uniaxial yield stress of polypropylene tapes with a draw ratio of 4 at 22°C for three different loading angles, (ϕ = 0°, 20°, 90°).

[Graph showing yield stress of PP tape versus strain rate for different loading angles on semi-logarithmic (a) and logarithmic scale (b). Squares are experimental results and lines are a guide to the eye.]

As expected for polymeric materials, the yield strength of the polypropylene tape increases with increasing strain rate. This strain rate dependence holds for all three orientations, although the increase of yield stress is most pronounced in the 0° orientation. Moreover, the slopes of the curves are significantly increasing with respect to the loading direction, see figure 3.1a, indicating that all orientations possess different deformation kinetics. However and more interestingly, the slopes of the strain rate dependence of the yield stress for different loading angles are similar, when plotted on a double logarithmic scale as shown in figure 3.1b. This indicates that the dependence of the yield stress on strain rate and orientation can be separated in a multiplicative way, i.e. that these effects are factorizable.
To investigate whether the observed factorizable behavior is also found with increasing draw ratio, further investigation is performed on tapes with different draw ratios ($\lambda = 1, 2, 4$ and $6$) and different off-axis loading angles ($\phi = 0^\circ, 20^\circ, 45^\circ$ and $90^\circ$). Figure 3.2 shows the uniaxial yield stress versus strain rate of polypropylene tapes with increasing draw ratios at $22^\circ C$ for two different orientations ($\phi = 0^\circ$ and $20^\circ$). The slopes of the lines in these figures are again similar for different draw ratio and loading angles. The strain rate dependencies for loading angles of $\phi = 45^\circ$ and $\phi = 90^\circ$ are not included in the figures since the values of the yield stress are only slightly increasing with draw ratio. Nevertheless, the slopes at these angles are still similar for increasing draw ratio. Furthermore, an increase in yield stress, for all strain rates, with increasing draw ratio is witnessed as expected for solid state drawn tapes.\textsuperscript{14}

The increase of yield stress is most pronounced in the drawing direction and decreases for increasing loading angles to the limit of $\phi = 90^\circ$, where the yield stresses are converging to the yield stress of the tape with draw ratio of $\lambda = 1$, i.e. to the mechanically isotropic polypropylene tape. Apparently, the draw ratio dependency is vanishing for increasing loading angle, so the draw ratio and loading angle are coupled parameters that together define the anisotropy of the material.

It is clear that the yield stress of anisotropic polypropylene depends on strain rate, draw ratio and loading angle. Moreover, on a double logarithmic scale the strain rate dependence on the yield stress shows factorizable behavior. Therefore, in the next section a viscoplastic model is proposed of the following form:

\begin{figure}[h]
\centering
\includegraphics[width=\textwidth]{figure3.2.png}
\caption{Yield stress of PP tape versus strain rate for increasingly oriented polypropylene with loading angles $\phi = 0^\circ$ (a) and $\phi = 20^\circ$ (b). Squares are experimental results and lines are a guide to the eye.}
\end{figure}
\[ \sigma_y = f(\dot{\varepsilon})g(\lambda,\varphi) \quad (3.1) \]

in which the function \( f(\dot{\varepsilon}) \) is accounting for the strain rate dependence and the function \( g(\lambda,\varphi) \) for the dependence on draw ratio and loading angle. The latter parameters are coupled as will be shown later.

### 3.2 Constitutive modeling

**Viscoplastic model**

A phenomenological anisotropic viscoplastic model for the rate-dependency of the yield stress is presented. In this model, the deformation rate tensor \( D \) is given by:

\[ D = \dot{\lambda}N, \quad (3.2) \]

with \( \dot{\lambda} \) a viscoplastic multiplier, which specifies the magnitude of plastic flow, while \( N \) determines the direction of plastic flow. The direction of the deformation rate is assumed to be given by an associated flow rule:

\[ N = \frac{\partial \sigma}{\partial \sigma}, \quad (3.3) \]

with \( \sigma \) an equivalent stress quantity, which will be defined as a function of the Cauchy stress tensor \( \sigma \).

The magnitude of plastic flow is defined such that the equivalent strain rate for anisotropic plasticity, which is defined as being work conjugate with the equivalent stress, is a function of the equivalent stress:

\[ \dot{\varepsilon} \equiv \frac{\sigma \cdot D \cdot \sigma}{\sigma} = \dot{\varepsilon}(\sigma), \quad (3.4) \]

As a result, the magnitude of plastic flow is written as:

\[ \dot{\lambda} = \frac{\sigma \cdot \dot{\varepsilon}(\sigma)}{\sigma \cdot N}, \quad (3.5) \]
Equivalent stress

The equivalent stress, in the previously described general viscoplastic model, is assumed to be based on Hill's anisotropic yield criterion:

\[
\bar{\sigma}_H^3 = H(\sigma_{11} - \sigma_{33})^3 + F(\sigma_{22} - \sigma_{33})^3 + G(\sigma_{33} - \sigma_{11})^3 + 2L\sigma_{23}^3 + 2M\sigma_{13}^3 + 2N\sigma_{12}^3,
\]

(3.6)

With

\[
F = \frac{1}{2} \left( \frac{1}{R_{22}^2} + \frac{1}{R_{33}^2} - \frac{1}{R_{11}^2} \right), \quad G = \frac{1}{2} \left( \frac{1}{R_{11}^2} + \frac{1}{R_{33}^2} - \frac{1}{R_{22}^2} \right), \quad H = \frac{1}{2} \left( \frac{1}{R_{11}^2} + \frac{1}{R_{22}^2} - \frac{1}{R_{33}^2} \right),
\]

(3.7)

\[
L = \frac{3}{2R_{23}^2}, \quad M = \frac{3}{2R_{13}^2}, \quad N = \frac{3}{2R_{12}^2}.
\]

(3.8)

where \( \sigma_{ij} \) are stress components with respect to a local material vector basis and the constants \( F, G, H, L, M \) and \( N \) define the anisotropic properties of the material.

The constants \( R_{11}, R_{22}, R_{33} \) are the ratios of the actual yield strength of the anisotropic material to the virtual bulk yield strength, \( \sigma_\gamma \). The constants \( R_{12}, R_{23}, R_{13} \) are the ratios of the yield strength values in shear to the shear yield strength \( \tau_\gamma \) of the virtual bulk material, with \( \tau_\gamma = \sigma_\gamma \sqrt{3} \). The actual yield strength in a certain loading direction is directly related to the draw ratio of the tape. Consequently, the constants \( R \) depend on the draw ratio. For the tensile tests performed in this study, the material 1-direction corresponds with the drawing direction and the 3-direction is the out-of-plane direction (see figure 2.2), for which a plane stress assumption is used:

\[
\sigma_{33} = \sigma_{23} = \sigma_{13} = 0
\]

(3.9)

Substitution into Hill's equivalent stress and using the anisotropic constants, leads to:

\[
\frac{\bar{\sigma}_H^3}{\sigma_\phi^3} = \frac{1}{R_{11}^2} \cos^4 \phi + \frac{1}{R_{22}^2} \sin^4 \phi - \left( \frac{1}{R_{11}^2} + \frac{1}{R_{22}^2} - \frac{1}{R_{33}^2} \right) \cos^2 \phi \sin^2 \phi + \frac{3}{R_{12}^2} \cos^2 \phi \sin^2 \phi \quad (3.10)
\]

Where \( \sigma_\phi \) is the stress applied at an angle \( \phi \) with respect to the drawing direction. Fibre symmetry, i.e. transversely isotropy, is assumed so that \( R_{22} = R_{33} \). The Hill equivalent stress in the drawing and the transverse direction, i.e. \( \phi = 20^\circ \) and \( \phi = 90^\circ \), is then determined only by the constants \( R_{11} \) and \( R_{22} \) respectively:
\[ \sigma_H^2 = \frac{1}{R_{11}} \sigma_0^2 \]  

(3.11)

and

\[ \sigma_H^2 = \frac{1}{R_{22}} \sigma_{90}^2 \]  

(3.12)

The observed macroscopically factorizable behavior is also found in Hill’s anisotropic criterion. Hill’s criterion is stated in a form which can be rewritten as:

\[ \log \sigma_\varphi = \log \sigma_H - \frac{1}{2} \log g (R_{ij} (\lambda), \varphi), \]  

(3.13)

with \( g (R_{ij} (\lambda), \varphi) \) the right-hand side of equation (3.10). The use of Hill’s criterion assumes that for each stress state corresponding to yielding, the equivalent stress is constant. The logarithm of the uniaxial yield stress in the direction given by \( \varphi \), is decomposed in the logarithm of the equivalent stress, i.e. a constant value, minus the logarithm of a function depending on the draw ratio, through the constants \( R_{ij} \), and the loading angle \( \varphi \). Therefore the orientation function \( g (\lambda, \varphi) \) shifts the equivalent stress to predict the yield stress for any loading angle.

Flow function

A flow function based on Eyring\textsuperscript{7,8} is proposed to determine the magnitude of plastic flow:

\[ \dot{\varepsilon} = \dot{\varepsilon}_0 \exp \left( \frac{-\Delta U}{RT} \left( \frac{\alpha V^*}{kT} \right) \right), \]  

(3.14)

Where \( V^* \) is the activation volume, determining the stress dependence, \( \Delta U \) the activation energy, determining the temperature dependence, \( \dot{\varepsilon}_0 \) is a rate constant, \( R \) is the universal gas constant, \( k \) is Boltzmann’s constant and \( T \) the absolute temperature.
A plastic deformation rate is obtained when a constant load is applied. Failure of a polymer sample will occur when a critical strain is exceeded. The time-to-failure is assumed to be given by:

\[ t_a = \frac{\varepsilon_{cr}}{\dot{\varepsilon}} \],

(3.15)

with \( \varepsilon_{cr} \) the critical equivalent strain.

### 3.3 Results

**Characterization**

The anisotropic constants of the proposed viscoplastic model are determined for different draw ratios. The orientation dependence of the yield stress is considered for only one (arbitrary) strain rate. Figure 3.3 shows the off-axis loading behavior for a number of draw ratios determined at a constant strain rate of \( 10^{-3} \text{ s}^{-1} \). Again, the yield stress dependence on draw ratio is clearly detectable.

![Figure 3.3: Off-axis yield stresses at a strain rate of \( 10^{-3} \text{ s}^{-1} \) for increasing draw ratio of PP tapes. Symbols are experimental results and lines are predictions of Hill's criterion.](image)

If solid state drawn tapes are loaded in the drawing direction, high yield stresses are found and the yield stress decreases with increasing loading angle. To use Hill's criterion, only the constants \( R_{11}, R_{22}=R_{33} \) and \( R_{12} \) have to be characterized as explained earlier. \( R_{11} \) and \( R_{22} \) are ratios of actual yield strength and the virtual
bulk strength. The actual yield strengths are given by the yield stresses measured at \( \phi = 0^\circ \) and \( \phi = 90^\circ \) respectively and for the virtual bulk strength, the yield stress at \( \phi = 90^\circ \) is chosen, which equals the yield stress of the undrawn material. Then, the constants \( R_{11} \) and \( R_{22} \) are determined with equations (3.11) and (3.12). The last remaining constant, \( R_{12} \), is determined from a fit to the experimental data with a least-squares minimization. The experimental off-axis yield stresses for different draw ratios are well described with Hill's anisotropic criterion and the parameters are presented in table 3.1.

<table>
<thead>
<tr>
<th>( \lambda )</th>
<th>( R_{11} ) [-]</th>
<th>( R_{12} ) [-]</th>
<th>( R_{22} ) [-]</th>
</tr>
</thead>
<tbody>
<tr>
<td>1</td>
<td>1.00</td>
<td>1.00</td>
<td>1.00</td>
</tr>
<tr>
<td>4</td>
<td>5.00</td>
<td>1.75</td>
<td>1.00</td>
</tr>
<tr>
<td>6</td>
<td>9.38</td>
<td>2.35</td>
<td>1.00</td>
</tr>
</tbody>
</table>

\[ \dot{\varepsilon}_0 \, [s^{-1}] \quad V^* \, [m^3] \quad \varepsilon_{cr} \, [-] \]

\[ 2.5 \cdot 10^{-10} \quad 2.5 \cdot 10^{-10} \quad 0.11 \]

Table 3.1: Model parameters determined with a least-squares minimization procedure.

The anisotropic viscoplastic model described in the previous section is used to predict the strain rate dependency of off-axis loading. Equations (3.2), (3.3) and (3.5) of the viscoplastic model are combined with Hill's anisotropic yield criterion and a simplification, excluding the temperature and pressure dependence, of the flow function. The remaining parameters which have to be characterized are the activation volume and rate constant in equation (3.14) and the critical equivalent strain in equation (3.15). Again, the yield stress data in the transverse direction, i.e. loading angle of \( \phi = 90^\circ \), is used to determine the parameters of the flow function and time to failure with a least-squared minimization procedure. All parameters are also presented in table 3.1.

Validation

It remains to be verified if the model correctly captures the strain rate dependency in different loading angles. Figure 3.3 shows the yield stress versus the off-axis loading angle with increasing strain rate on a semi-logarithmic scale. Over the whole range of angles the yield stress for different strain rates differs by a constant value. Therefore, the dependence of the yield stress on the loading angle appears to be independent of the applied strain rate. The predicted curves are
constructed with identical parameters for $\lambda = 6$, given in table 3.2, for the orientation function for all strain rates.

![Graph](image)

**Figure 3.3**: Off-axis yield stresses for increasing strain rate for oriented polypropylene with draw ratio of four. Symbols are experimental results and lines are predictions of Hill’s criterion.

Finally, the viscoplastic model is used to predict the yield stress in different loading directions over a wide range of strain rates for two tapes with different draw ratios of $\lambda = 4$ and $\lambda = 6$ as shown in figure 3.4. The fitting procedure is now shortly summarized; The orientation and subsequently the draw ratio dependence is determined for an arbitrary strain rate, in this case $10^{-3}$ s$^{-1}$. The 90° orientation is used as a reference to determine the strain rate dependence and the critical equivalent strain. Finally, by factorizing the reference state different orientations are predicted.

For all loading angles, the prediction of the strain rate dependency is in good agreement with the experimental data and also time to failure is correctly predicted for both draw ratios with only one value of the critical equivalent strain for all loading angles, see figure 3.4.
Figure 3.4: Yield stress and time-to-failure for different loading angles at room temperature with $\lambda = 4$ (a and b) and $\lambda = 6$ (c and d). Squares are experimental results and lines are model predictions.
Chapter 4

Micro structural aspects of plastic flow

In the previous chapter a model is presented that captures the macroscopic anisotropic plastic flow response of oriented semi-crystalline polymers. The micro structural processes that control the mechanical behavior, i.e. yielding or tensile strength, of the crystalline phase in semi-crystalline polymers are associated with crystallographic slip. This chapter focuses on these micro structural aspects of plastic deformation.

4.1 Molecular origin of deformation

In the case of polymer glasses, typical features during deformation are yield, strain softening (the decrease in stress that is observed after passing the yield point) and strain hardening at large deformations. The yield stress is the onset of main-chain segmental motion induced by stress. At yield the deformation becomes irrecoverable since stress-induced plastic flow sets in, leading to a structural evolution which reduces the material’s resistance to plastic flow; strain softening. The yield stress is influenced by the rate of deformation and the time upon ageing. In figure 4.1a it is shown that the yield stress increases with increasing strain rate. Strain softening is closely related to the occurrence of physical ageing. This is demonstrated in figure 4.1b that compares the intrinsic response of a slow and fast cooled polycarbonate sample. Annealing results in an increase of yield stress, but upon plastic deformation the differences between both curves disappear and they fully coincide after strain softening. Apparently all influence of the thermal history is erased due the softening. Strain hardening is generally interpreted as the result of a stress contribution of the orienting molecular network and therefore the stress increases at large deformations.
The deformation behavior of semi-crystalline polymers generally possesses the same features as amorphous polymers, such as yielding, strain softening and strain hardening, see figure 4.2a. However, the crystallinity and subsequently the lamellar thickness are significantly influencing the deformation behavior. The initial deformation of semi-crystalline polymers is explained by interlamellar strain of the amorphous part and at yield, break up of lamellae occurs by slip in the crystalline lamellae. A higher degree of crystallinity, or better an increase of lamellar thickness, leads to an increase of yield stress. However, the strain hardening of semi-crystalline polymers is not affected by the degree of crystallinity.
For polymer materials, it is generally observed that the time-dependent material behavior is caused by molecular transitions and these transitions are activated by temperature. The deformation kinetics, for a wide range of strain rates and elevated temperatures, can reveal the contribution of the amorphous and crystalline phase, respectively. As an example, figure 4.3a shows the yield stress of isotropic polypropylene as a function of strain rate for several elevated temperatures. The strain rate dependence of the yield stress of isotropic polypropylene displays a clear change in slope, which is shown to be related to the onset of a stress contribution of an additional molecular process. Usually, these mechanisms are modeled as two parallel dashpots, implying that the stress contributions are additive, see figure 4.3b. The $\alpha$-process (at high temperatures and/or low strain rates) is attributed to the crystalline phase and the $\beta$-process is related to the glass transition, i.e. the amorphous phase.\footnote{3}

The origin of the crystalline contribution to the yield stress is found in the micro structural processes that control the plastic deformation of the crystalline phase and these processes are associated with crystallographic slip. This plastic deformation process and the influence of lamellar thickness on the yield stress are discussed in the next section.

Two questions remain after this summary of the molecular origin of deformation. First, do kinetics of slip (i.e. activation energy, activation volume and rate constant) differ for the different slip systems? Secondly, how do the kinetics depend on the lamellar thickness?

\begin{figure}[h]
\centering
\includegraphics[width=\textwidth]{figure4.3.png}
\caption{Schematic representation of the decomposition of the strain rate dependence of the yield stress into two separate molecular contributions (a) and the ratio of yield stress to temperature as a function of strain rate for polypropylene reproduced from Klompen.\footnote{12} (b)}
\end{figure}
4.2 Crystalline deformation processes

The plastic deformation of semi-crystalline polymers is generally expected to be crystallographic in nature. The micro structural processes that control the mechanical behavior, i.e. yield or tensile strength, of the crystalline phase in semi-crystalline polymers are primarily associated with crystallographic slip and secondarily with mechanical twinning or stress-induced martensic phase transition. The latter two processes are left out of consideration since crystallographic slip is considered to be of most importance.

The micro structure of the oriented polypropylene tape is characterized by lamellar stacks, consisting of layers of crystalline and amorphous regions. The crystalline lamellae are oriented with their chain axis, i.e. the c-axis, almost parallel to the drawing direction, see figure 4.4a. It is worth mentioning that the shown WAXS pattern is measured for a tape with draw ratio of 7, nevertheless it is assumed that the oriented structure of a tape with draw ratio of 6 is similar. Since the chain axis is almost parallel to the drawing direction, a random orientation of the a- an b-direction around the c-axis is assumed.

In this study it is assumed that only α-crystals are present, since there is no evidence in the WAXS profile for β- or γ-crystals. Neither a characteristic β-crystal peak at $2\theta \approx 16^\circ$, nor a γ-peak at $2\theta \approx 20^\circ$ is observed in figure 4.4b. The polypropylene crystal lattice of the α-crystals is monoclinic with lattice parameters $a = 6.66$ Å, $b = 20.98$ Å, $c = 6.65$ Å and $\beta = 99.3^\circ$.

In this work, eight slip systems are taken into consideration, listed in table 4.1, which can occur in polypropylene α-crystals either with the slip direction along the chain axis, i.e. chain slip, or perpendicular to the chain axis, i.e. transverse slip.
The possible slip systems for α-crystals in polypropylene are depicted in figure 4.5. Crystallographic slip is dominant in plastic deformation and therefore many theoretical estimations and experimental predictions are reported in literature. The easiest overall chain slip system in polypropylene crystals is (010)[001], while the (100)[001] and {110}[001] systems have a higher critical resolved shear stress. Furthermore, a few transverse slip systems are reported and are given by the (100)[010], (010)[100] and {110}<1\overline{1}0> slip systems. It is not stated which slip system is dominant.

<table>
<thead>
<tr>
<th>Slip system</th>
<th></th>
</tr>
</thead>
<tbody>
<tr>
<td>Chain slip</td>
<td>(100)[001]</td>
</tr>
<tr>
<td></td>
<td>(010)[001]</td>
</tr>
<tr>
<td></td>
<td>{110}[001]</td>
</tr>
<tr>
<td>Transverse slip</td>
<td>(100)[010]</td>
</tr>
<tr>
<td></td>
<td>(010)[100]</td>
</tr>
<tr>
<td></td>
<td>{110}&lt;1\overline{1}0&gt;</td>
</tr>
</tbody>
</table>

Table 4.1: Eight slip systems of polypropylene taken from G'Sell et al.\textsuperscript{30}

\textsuperscript{1} Slip systems are generally represented by the Miller index notation system of the slip plane, in this case (010), followed by those of the slip direction, here [001]. Using this notation, the six displayed slip systems in table 4.1 represent eight physically distinct slip systems.
Figure 4.5: Isometric and topview of the monoclinic unit cell of the α-crystal of polypropylene constructed of figures adapted from van der Burgt and Lotz et al. Possible slip systems are indicated by black arrows within the open planes.
The micro structure of the polypropylene tape is highly oriented and therefore the chain axis is aligned almost parallel with the drawing direction. Here a misalignment of $\xi = 2^\circ$ is assumed with $\xi$ the angle between the c-axis and the drawing direction. The a- and b-directions of the unit cell are randomly distributed around the c-axis. For all different loading directions, the absolute value of the Schmid factor, defined as the ratio of the resolved shear stress in a slip system and the applied uniaxial stress as shown in figure 4.6, is calculated to establish which slip systems are most likely activated. It is worth mentioning that slip occurs only when the resolved shear stress on the slip plane reaches a critical value, known as the critical resolved shear stress. Therefore, only in combination with these critical values, the values of the Schmid factor indicate which slip system are activated for different loading directions.

![Figure 4.6: Definition of a slip system: slip plane and slip direction.](image)

For deformation in the drawing direction, practically no transverse slip is possible since the small misalignment and the angle $\beta$ between the a- and c-direction. Consequently, chain slip is the main deformation mechanism in the drawing direction. Since the chain axis in the crystals are aligned parallel to the drawing direction, all chain slip systems are unfavorably oriented, resulting in very low values of the Schmid factor, see figure 4.7a. The chain axis itself is inextensible by the mechanism of crystallographic slip. However, the [001] slip direction is slightly activated due to the small angle between the chain axis in the crystalline lamellae and the loading axis. The plane of possible slip depends on the direction of misalignment.

For deformation in the transverse direction, the load and the chain direction in the crystalline lamellae are perpendicular. Therefore, chain slip systems will be inactive, but transverse slip systems may be active. The theoretical predictions of the Schmid factor indicate that all three transverse slip systems may be active, see
figure 4.7c. The slip systems \{110\}<1\overline{1}0> are most favorably oriented in the transverse direction to accommodate crystallographic slip.

In the previous deformation modes, i.e. deformation in the drawing and transverse direction, only one slip mechanism is dominant; chain or transverse slip respectively. Both mechanisms are activated in the 20° deformation mode as can be seen in figure 4.7b and 4.7d. Chain slip is significantly favored and transverse slip is less pronounced compared to the 90° orientation. For the chain slip mechanism, the (010)[001] slip system is most favorably oriented, however the differences with \{110\}[001] are very small. It is not possible to determine the most likely activated slip system for the transverse slip mechanism.

It is now clear that the crystalline phase, assigned to the α-process in semi-crystalline polymers, can deform by chain or transverse slip mechanisms and these slip mechanisms are single or simultaneously activated in different loading conditions.
orientations. The important question remains if the activated slip systems possess similar deformation kinetics. In the previous chapter, where only the macroscopic behavior was captured, the deformation kinetics for all deformation modes was similar. However, it is still logical to assume that different slip systems possess different kinetics. To investigate this, the experimental range will be increased by increasing the temperature.

The anisotropy of a tape can be described by the anisotropic constants of Hill’s yield criterion which refer to the transverse direction. Furthermore, the transverse direction has similar mechanical properties as the isotropic tape. So, it should be logical to first explore the temperature dependence in the transverse direction and then to translate to different loading directions.

Since the $\varphi = 90^\circ$ orientation and the isotropic state are mechanically similar, it is expected that a description of two Eyring processes should be sufficient to obtain good yield stress predictions. The lower temperature transition $\beta$ is the glass transition and is attributed to the amorphous phase and related to main-chain segmental motion. The $\alpha$-process is attributed to processes within the crystalline phase. Figure 4.8b schematically represents those two processes and both contributions are additive in stress. The temperature and strain rate dependence of polypropylene tapes with a draw ratio of $\lambda = 6$ is shown in figure 4.8a for several temperatures in the range of 22-60°C. In figure 4.8a it is shown that the obtained fit for $\varphi = 90^\circ$ orientation is sufficient to describe the yield stress data. The parameters used are presented in table 4.2. The Schmid factors in the previous section showed that only transverse slip can occur in the transverse loading direction. Hence, the plastic deformation in the $\alpha$-regime is solely accommodated by transverse slip mechanisms.

Figure 4.8: Yield stress as function of strain rate for oriented polypropylene with draw ratio $\lambda = 6$ for a loading angle of $\varphi = 90^\circ$ (a), with respect to the drawing direction at elevated temperature, symbols represent experimental results and lines are Eyring fits and a schematic representations of the decomposition of the strain rate dependence of the yield stress into two separate molecular contributions in parallel (b).
Table 4.2: Parameters for two Eyring processes obtained from yield data in figure 4.8-10.

<table>
<thead>
<tr>
<th>Orientation</th>
<th>φ = 20° / 90°</th>
<th>φ = 0°</th>
</tr>
</thead>
<tbody>
<tr>
<td>Process</td>
<td>α</td>
<td>β</td>
</tr>
<tr>
<td>(\Delta U) [kJ.mol(^{-1})]</td>
<td>394.0</td>
<td>162.72</td>
</tr>
<tr>
<td>(V^*) [m(^3)]</td>
<td>7.0 (\times) 10(^{-27})</td>
<td>2.4 (\times) 10(^{-27})</td>
</tr>
<tr>
<td>(\dot{\gamma}_0) [s(^{-1})]</td>
<td>7.7 (\times) 10(^{29})</td>
<td>4.2 (\times) 10(^{23})</td>
</tr>
</tbody>
</table>

The temperature dependence is now established for the transverse direction and the yield stress data at \(φ = 0°\) and \(φ = 20°\) are predicted by factorization as done in chapter 1. For \(φ = 20°\), the prediction of the yield stress presented in figure 4.9b is considered satisfactory. However, for an orientation of \(φ = 0°\) the yield stress is overestimated at low strain rates and elevated temperatures, see figure 4.9a.

![Figure 4.9](image)

**Figure 4.9:** Yield stress as function of strain rate for oriented polypropylene with draw ratio \(\lambda = 6\) for different loading angles, \(φ = 0°\) (a), \(φ = 20°\) (b), with respect to the drawing direction at elevated temperature. Symbols represent experimental results and lines are Eyring fits.

In the drawing direction, chain slip mechanisms are dominant, i.e. the Schmid factor values of transverse slip are approximately zero, and consequently factorizing the reference state is incorrect since in that state only transverse slip mechanisms are active. Normally, the contribution of the \(α\)-process is shown as a leveling off of stress at low strain rate and/or high temperatures. In this case, the inverse is observed in the experimental data; at low strain rate and/or high temperatures the stress is still decreasing. It is proposed that this can be modeled as two dashpots in series and subsequently the strain rate in both processes are
additive. This is schematically represented in figure 4.10b. This can be understood since the micro structure of the oriented tape is highly oriented, which can be interpreted as blocks of crystalline lamellae and amorphous regions in series with respect to the drawing direction.

The strain rate dependence of polypropylene tapes with draw ratio $\lambda = 6$, loaded in the drawing direction, at elevated temperatures is shown in figure 4.10a. The solid lines are yield stress predictions with two dashpots in series as explained and the used parameters are also given in table 4.2. In the case of loading in the drawing direction the $\alpha$-process is related to chain slip mechanisms.

![Figure 4.10](a) Yield stress as a function of strain rate for oriented polypropylene with draw ratio $\lambda = 6$ loaded in the drawing direction at different elevated temperatures. Symbols represent experimental results and lines are predictions.

(b)

Figure 4.10: Yield stress as a function of strain rate for oriented polypropylene with draw ratio $\lambda = 6$ loaded in the drawing direction at different elevated temperatures. Symbols represent experimental results and lines are predictions.

Different values of the activation energy are found for both slip mechanisms by fitting the yield stress data. In addition to the dependence of the activated slip mechanism for different loading directions, it can be concluded that slip mechanisms possess different deformation kinetics.

4.3 Influence of lamellar thickness on deformation kinetics

Several theories have been developed to describe the yielding of the crystalline phase. The proposed model of Young describes the yield stress dependence on lamellar thickness. This model assumes that yield involves the thermal activation of screw dislocations with the burgers vector parallel to the chain axis. Analysis of Young's slip model leads to the conclusion that crystallinity is not the critical parameter but lamellar thickness is the controlling parameter in the
activation of yield. The latter statement is supported by investigations of Schrauwen et al. and Galeski. Both found a proportionality between lamellar thickness and yield stress. This proportionality is only valid for lamellar thicknesses in the range of 10 to 40nm. Above a thickness of 40nm, only obtained under specific conditions such as high pressure, the yield stress saturates, see figure 4.11.

![Figure 4.11: Relationship between lamellar thickness and yield stress of PE for different strain rates, reproduced from Galeski.](image)

The influence of lamellar thickness on the yield stress is investigated for isotropic polyethylene. Different crystallization procedures are applied to obtain different crystallinities and subsequently lamellar thicknesses. The obtained samples are uniaxially tested at different strain rates to detect the deformation kinetics on lamellar thickness.

An example of the influence of lamellar thickness on yield stress at a single strain rate is provided in the work of Schrauwen et al. The lamellar thickness for different polymers (PE, PP, PET) is controlled by the processing conditions via quenching, annealing and slowly cooling of compression molded sheets. In this work, the alteration of crystallinity and subsequent determination of crystallinity, long spacing and lamellar thickness is repeated for two grades of polyethylene. Via quenching and annealing or slow cooling of the samples crystallinities and lamellar thicknesses in the range of 67–77% and 17–28nm are obtained respectively, presented in table 4.3.
Material Procedure $\chi$ [%] $D$ [nm] $\delta$ [nm]

<table>
<thead>
<tr>
<th>Material</th>
<th>Procedure</th>
<th>$\chi$ [%]</th>
<th>$D$ [nm]</th>
<th>$\delta$ [nm]</th>
</tr>
</thead>
<tbody>
<tr>
<td>PE 8621</td>
<td>Q15</td>
<td>70.93</td>
<td>28.03</td>
<td>19.88</td>
</tr>
<tr>
<td></td>
<td>IC120</td>
<td>73.88</td>
<td>31.43</td>
<td>23.22</td>
</tr>
<tr>
<td></td>
<td>SC</td>
<td>76.57</td>
<td>37.47</td>
<td>28.69</td>
</tr>
<tr>
<td>PE 9089S</td>
<td>Q15</td>
<td>67.80</td>
<td>25.41</td>
<td>17.23</td>
</tr>
<tr>
<td></td>
<td>IC120</td>
<td>71.20</td>
<td>28.64</td>
<td>20.39</td>
</tr>
<tr>
<td></td>
<td>SC</td>
<td>76.55</td>
<td>32.58</td>
<td>24.94</td>
</tr>
</tbody>
</table>

Table 4.3: Degree of crystallinity $\chi$, long period $D$ and lamellar thickness $\delta$ of polyethylene samples with different crystallization procedures applied.

Figure 4.12: Yield stress of two grades of polyethylene PE 8621 (open symbols) and PE 9089S (filled symbols) versus lamellar thickness determined at a strain rate of $10^{-3}$ s$^{-1}$. The lines are a guide to the eye.

Figure 4.12 gives the yield stress as a function of lamellar thickness for a strain rate of $10^{-3}$ s$^{-1}$. Independent of the molecular weight, i.e. different grades, of the polyethylene samples a proportionality is found between lamellar thickness and yield stress for both polyethylene grades at one strain rate.

The strain rate dependency of the yield stress is investigated for samples of polyethylene with different lamellar thicknesses at room temperature and a strain rate range of $10^{-2}$-$10^{-4}$ s$^{-1}$. In figure 4.13a an increase of yield stress with increasing strain rate is shown for samples with different crystallization procedures. Even more interesting, the strain rate dependence is similar for different lamellar thicknesses.
Figure 4.13: Yield stress data for three different crystallization procedures, i.e. quenching, annealing and slow cooling, at a strain rate of $10^{-3}$ s$^{-1}$ and at room temperature. Lines are a guide to the eye to indicate similar strain rate dependence (a) and a schematical representation of the influence of increasing lamellar thickness on the strain rate dependence of the yield stress (b).

The similar strain rate dependence for samples with different crystallization procedures applied suggests that a mastercurve for lamellar thickness can be constructed by horizontal shifting only. A state parameter $S_\delta$ is proposed to construct the mastercurve. In this approach $S_\delta$ decreases the pre-exponential factor of the flow function, given in eq. (4.13). This is schematically demonstrated in figure 4.13b, which shows the strain rate dependent yield stress resulting from eq. (3.14). The value of this state parameter indicates how strongly the present situation of the polymer deviates from the reference state. Plotting the yield stress versus the logarithm of strain rate, an increase of $S_\delta$ basically shifts the curve horizontally towards lower strain rates. At a constant strain rate this results in an increase of the yield stress compared to that of the reference state. The desired lamellar thickness dependence is added by introducing the state parameter $S_\delta$ into eq. (3.14):

$$
\tilde{\varepsilon} = \dot{\varepsilon}_{0,\delta} \exp(-S_\delta) \exp\left(-\frac{\Delta U}{kT}\right) \sinh\left(\frac{\sigma V^*}{kT}\right), \quad (4.1)
$$

where $\dot{\varepsilon}_{0,\delta}$ is the reference rate constant. The parameter $S_\delta$ is used as a fitting parameter rather than a true physical parameter.
The state parameter $S_{\delta}$ used to construct the mastercurve is plotted against lamellar thickness in figure 4.14. A linear relation is observed for the state parameter with increasing lamellar thickness.

For polymers however, the temperature dependence is sometimes even more pronounced than the strain rate dependence. The polyethylene samples obtained with different crystallization procedures are uniaxially tested at several elevated temperatures in the range of 22°-60°C and strain rates of $10^{-2}$-$10^{-5}$ s$^{-1}$. The yield data are described with a single thermally activated Eyring process for the quenched samples. The quenched state is taken as a reference for the mastercurves for lamellar thickness. To construct the mastercurve, the activation volume, activation energy, rate constants and the state parameter are determined with a least-squares minimization procedure. The parameters obtained are presented in table 4.4.

<table>
<thead>
<tr>
<th>Material</th>
<th>$\Delta U$ [kJ.mol$^{-1}$]</th>
<th>$V^*$ [$m^3$]</th>
<th>$\dot{\varepsilon}_{0,m} [s^{-1}]$</th>
</tr>
</thead>
<tbody>
<tr>
<td>PE 8621</td>
<td>203.75</td>
<td>$4.2 \cdot 10^{-27}$</td>
<td>$3.0 \cdot 10^{27}$</td>
</tr>
<tr>
<td>Q15</td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>IC120</td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>SC</td>
<td></td>
<td></td>
<td></td>
</tr>
</tbody>
</table>

<table>
<thead>
<tr>
<th>$S_{\delta}$ [-]</th>
</tr>
</thead>
<tbody>
<tr>
<td>Q15</td>
</tr>
<tr>
<td>IC120</td>
</tr>
<tr>
<td>SC</td>
</tr>
</tbody>
</table>

Table 4.4: Fitted parameters to construct the mastercurve, with quenched polyethylene as a reference state.
The possibility of constructing a mastercurve by horizontal shifting indicates strain rate-lamellar thickness superposition. This implies that the strain rate decreases by the same factor as the lamellar thickness increases. Mastercurves are constructed with constant state parameters for different temperatures. It is clearly seen in figure 4.15b that the data at temperatures of 22°C and 40°C can be represented by a single thermally activated Eyring process. For an elevated temperature of 60°C however, the data is under predicted. Apparently, another molecular process is contributing. Still, it is clearly observed that the strain rate-lamellar thickness superposition for a single activated process holds for all temperatures. Furthermore, the yield data also suggest that this superposition holds for the second process at high temperatures and low strain rates.

To validate the linear proportionality between the state parameter and lamellar thickness, the predicted values of $S_δ$ are used to fit the experimental yield data of samples with different lamellar thicknesses at elevated temperatures using eq. (4.1). In figure 4.15 it is clear that the predictions of the yield stress are in good agreement with the experimental data.

![Figure 4.15](image-url)

**Figure 4.15**: Yield stress as function of strain rate for PE samples with different crystallization procedures applied at different temperatures. Symbols are experimental data and solid lines are fits using eq. (4.1) (a). Mastercurves constructed from the data with the quenched state for each temperature used as reference (b).
Conclusions

The basis of this study consists of the mechanical characterization of the off-axis material behavior of solid state oriented polypropylene, using yield stress results from uniaxial tensile experiments at different strain rates and time to failure data, both at different loading angles with respect to the drawing direction. The off-axis behavior is well described with Hill’s anisotropic yield criterion. With the assumption of a plane-stress situation only three parameters, i.e. the anisotropic constants, can describe the off-axis behavior. Herein, the constant $R_{22}$ is independent of draw ratio and determines the $\theta = 90^\circ$ orientation, which can be regarded as similar to the mechanically isotropic state. The remaining two constants are dependent on draw ratio and increase with increasing draw ratio. It is shown that both draw ratio and orientation are factorizable with respect to the $\theta = 90^\circ$ orientation at room temperature. Consequently, once the strain rate dependence of the transverse direction is characterized with the proposed viscoplastic model, the yield stress and time to failure can be predicted at different orientations. The critical equivalent strain, necessary to predict the time to failure, is independent of orientation and draw ratio.

Polyethylene samples are obtained with different crystallization procedures, i.e. quenching, subsequently annealing and slow cooling, applied. These procedures have a significant effect on the degree of crystallinity and subsequently on lamellar thickness which resulted in crystallinities and lamellar thicknesses in the range of 67-77% and 17-28nm, respectively. The yield stress at which slip of crystalline lamellae is reached is shown to be dependent on the lamellar thickness. Furthermore, the strain rate dependence on the yield stress at room temperature is similar for increasing lamellar thickness. The activation of yielding is described with a single Eyring process with the introduction of a state parameter $S$. Using this parameter, a mastercurve can be constructed by horizontal shifting only of the yield stress for samples with increasing lamellar thickness. The horizontal shifting implies strain rate-lamellar thickness superposition. This superposition is shown for a single Eyring process, however, also for the second molecular process this superposition probably holds.

The anisotropy of a tape can be described by the anisotropic constants of Hill’s yield criterion which refer to the transverse direction. The transverse direction has similar mechanical properties as the isotropic tape. The temperature dependence in the isotropic state is explored and then translated to different loading directions. The strain rate dependence in the transverse direction of oriented polypropylene tape is well described with two thermally activated Eyring processes. By factorization of the transverse direction, the strain rate dependence for $\varphi=20^\circ$
loading is satisfactorily predicted indicating similar slip kinetics. However, this factorization does not hold for $\varphi=0^\circ$ loading. Apparently, different slip kinetics are involved in the drawing direction. The deformation of the crystalline phase in the drawing or transverse direction are given respectively by chain or transverse slip solely. Still, further investigation is recommended on the actual activation of slip systems, i.e. when the critical resolved shear stress is exceeded, and which slip systems are activated and dominant.
Bibliography


