Micromechanical Modeling of the Tensile Behavior of Oriented Polyethylene


Dutch Polymer Institute, Eindhoven University of Technology, Department of Mechanical Engineering, PO Box 513, 5600 MB Eindhoven, The Netherlands

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ABSTRACT: The stacked lamellar morphology commonly found in extruded semicrystalline materials has a strong influence on the flow direction, with respect to the loading direction, and on the stability and localization phenomena in tensile experiments. A multiscale numerical model was used to simulate the effect on the macroscopic behavior of a stacked lamellar microstructure. The model established a link between the microscopic, the mesoscopic, and the macroscopic levels. The constitutive properties of the material were identified for the crystallographic and amorphous domains. The average fields of an aggregate of individual phases, having preferential orientations, formed the constitutive behavior of the extruded material. The microscopic morphology of the extruded high-density polyethylene is based on wide-angle X-ray diffraction experiments. The macrostructure was described by a finite element model. The microstructure-induced deformation hardening in the extrusion direction was found to stabilize the macrostructure when it was loaded in the flow direction. © 2004 Wiley Periodicals, Inc.

Keywords: extrusion; microstructure; orientation; polyethylene; semicrystalline

INTRODUCTION

The mechanical performance of semicrystalline polymeric materials, as characterized by elasto-viscoplastic deformation, is strongly dependent on the underlying microstructure. Semicrystalline polymers consist of both amorphous and crystalline domains. The elastic and the viscoplastic behavior depend on the percentage of crystallinity, the initial crystallographic and morphological texture, and the evolution of this microstructure with deformation. In recent years, much experimental work and many modeling studies have focused on understanding viscoplastic behavior and the evolution of the texture of semicrystalline polymers. A micromechanically-based model for the constitutive behavior of semicrystalline polymeric material was presented in Van Dommelen et al. The model accounts for both crystallographic and morphological texture, the latter corresponding to the orientation distribution of the lamellar interface normals. In Van Dommelen et al. a three-level modeling approach was used to study intraspherulitic deformation and stresses for the semicrystalline polyethylene. The current study builds on the recently developed models with this micromechanical framework for the simulation of the mechanical behavior of oriented high-density polyethylene (HDPE).

For melt-extruded material, a stacked lamellar morphology is often observed. For this material, the mechanical response will depend on the
direction of loading with respect to the flow direction, as is illustrated by the images in Figure 1. This figure shows three specimens that were prepared from extruded material at different angles with respect to the extrusion direction. When loaded in the extrusion direction, the deformation remained homogeneous throughout the entire sample [see Fig. 1(a)], whereas when loaded perpendicular to the extrusion direction, a neck was formed [see Fig. 1(c)]. The neck formation mechanism in semicrystalline polymers has been related to the unfolding of chains in the crystalline domains.20–24

In this study, a multiscale numerical model was used to investigate the deformation mechanics of oriented polyethylene. For this purpose, a distinction between three different scales was made, as is schematically shown in Figure 2. The macroscopic structure, for instance, a tensile bar, was modeled by finite elements. In each material point, referred to as the mesoscopic level, an assembly of lamellae was assumed. The mechanical behavior of this assembly was modeled by a polycrystalline composite inclusion model.15 The basic element of this model was a two-phase layered structure of a crystalline and an amorphous domain. Constitutive properties were assigned to the components at this microscopic level. Crystallographic and morphological texture was represented by the orientations of the crystalline domains and the amorphous/crystalline interfaces, respectively. The microstructural behavior was linked to the mesoscopic scale by an interaction law and to the macroscopic scale by finite element modeling.

The mechanical response of oriented semicrystalline polymers was studied, depending on the initial microstructure. The anisotropic mechanical behavior of HDPE with a stacked lamellar morphology was investigated. Various assumed microstructures, for both twisted and untwisted material, were used and were based on wide-angle X-ray diffraction (WAXD) measurements of extruded HDPE. Finite element simulations of tensile experiments were used to study the localization behavior, depending on the orientation of the extrusion direction relative to the loading direction.

EXPERIMENTAL

The HDPE used in this study was supplied by DSM (The Netherlands; Stamylan HD 8621, $M_w = 210,000$ g/mol, $M_n = 7000$ g/mol). Oriented polymer sheet samples were manufactured by film/sheet extrusion.25 A Collin Teach-Line single screw extruder of 20 mm diameter, operating at 1800 rpm, was used to establish a constant flow of 41.5 g/min. The flow was stabilized in a channel of $15 \times 15$ mm$^2$ over a length of approximately 250 mm before entering the sheet extrusion die. The temperature of the extruder barrel, the stabilization channel, and the die were all set to 180 °C. The exit cross-section of the sheet die was set to $2.1 \times 100$ mm$^2$ and the distance from the die to a set of chill rolls was approximately 150 mm. The chill rolls were water-cooled to 20 °C and the thickness of the final sheet was controlled by the pressure ($p$) and the speed ($\Omega$) of the rolls. During sheet extrusion, part of the polymer was oriented
by the flow inside the extrusion die, however, most orientation was due to drawing of the melt by the chill rolls. The speed of the rolls was varied and the draw ratio (DR) of the melt between the die and the chill rolls was determined as the ratio of the initial and the final cross-section of the sheet. The chill roll speed, the pressure, and the draw ratio of different extruded samples are listed in Table 1.

### Table 1. Extrusion Molding Conditions

<table>
<thead>
<tr>
<th>Sample code</th>
<th>Ω [rpm]</th>
<th>p [bar]</th>
<th>DR [-]</th>
</tr>
</thead>
<tbody>
<tr>
<td>PE-A</td>
<td>0.8</td>
<td>3</td>
<td>5.3</td>
</tr>
<tr>
<td>PE-B</td>
<td>2.4</td>
<td>5</td>
<td>17.5</td>
</tr>
</tbody>
</table>

Orientation Measurement

Flow-induced oriented structures in the extruded samples were analyzed by recording 2D WAXD film patterns. WAXD experiments were performed at the European Synchrotron Radiation Facilities (ESRF) in Grenoble, France, at the micro-focus beam line (ID13). Because the initial beam size was 5 μm and the collimator-to-sample distance was about 3 mm, the resulting beam size transmitting through the sample (TD direction) was less than 10 μm, enabling the examination of different structures over the thickness of the specimens. Samples 1 mm in width were cut from the middle of the specimens and WAXD patterns were recorded with a 2D-MAR CCD detector (resolution: 2048 × 2048) scanning the sample with a moving stage over the thickness of the specimen (TD direction). The exposure time was on the order of 1–5 s, with an X-ray wavelength of 0.975 Å and a sample-to-detector distance of 150 mm. Scanning was performed in steps of 10 μm over the full thickness of the extruded sheets. The resulting X-ray diffraction patterns of the extruded polyethylene samples showed an approximately uniform distribution of orientation over the sample thickness. Pole figures were constructed to characterize the three-dimensional crystallographic orientation distribution. For this purpose, samples were cut from the extruded sheets with a width equal to the sheet thickness, resulting in square samples and reducing intensity differences because of path length variations. These samples were positioned on a goniometer with the rotation axis in the middle of the ND–TD plane of the sample (see Fig. 3) and rotated around the machine direction (MD) over 90° only (because of sample symmetry), collecting WAXD patterns every 5°. Intensities of these collected patterns were taken every 5° over the azimuth angle of the 110, 200, and 020 Debye rings and used as input data to create full pole figures in the software program BEARTEX.

The resulting orientation distribution is shown in equal area pole figures for the principal lattice planes for two different extrusion conditions (Fig. 4). In these figures, MD corresponds to the extrusion direction and ND is the normal direction of the extruded sheet. For the material denoted by PE-A (draw ratio 5.3), the molecular chains (i.e., the c-axes) were oriented preferentially in the TD–MD plane, with a small maximum in the extrusion direction. The b-axis, corresponding to the lamellar growth direction, is oriented preferentially perpendicular to the extrusion direction. These observations indicate a row-like structure with some degree of twisting. The sample denoted by PE-B was obtained with a higher draw ratio (17.5). The strong orientation of molecular chains in the extrusion direction and the a-axes and b-axes being oriented in approximately random directions in the ND–TD plane indicate a microstructure of untwisted lamellae, perpendicular to the flow direction.

MECHANICAL RESPONSE

For tensile testing, specimens were machined from the oriented sheets with sample code PE-B for different angles (φ) with respect to the extrusion direction. The geometry of the specimens is shown in Figure 5(a). Tensile tests on the machined specimens were performed on a Zwick
Z010 tensile testing machine, at a constant speed of $3.75 \times 10^{-3}$ mm/s with an initial clamp-to-clamp distance of 15 mm. Engineering stress–strain curves were constructed from the measured force, clamp displacement, and initial cross-section of the specimens, and are shown in Figure 5(b). When loaded in the extrusion direction, continuing post-yield strain hardening is observed, which stabilizes the sample during the deformation process. In the specimen that is machined perpendicular to the extrusion direction, a neck is formed and, consequently, macroscopic strain softening is found. For the sample that is machined at an angle of 45° with the extrusion direction, no sharp maximum of the stress–strain curve is observed.

**MODEL DESCRIPTION**

The constitutive behavior of semicrystalline material was modeled by an aggregate of two-phase composite inclusions. This composite inclusion model, which is discussed in detail in ref. 15, is concisely summarized in this section.

Each inclusion consisted of a crystalline and an amorphous phase. A microstructural elasto–viscoplastic constitutive model was defined for both the crystalline and the amorphous phase. In the following sections, the constitutive models for the elastic and plastic parts of each phase are discussed. These models were based on a multiplicative decomposition of the plastic and elastic components of each phase.\textsuperscript{29}
Crystalline Phase

The crystalline domain of polymeric material consists of regularly ordered molecular chains. The crystal structure results in (1) anisotropic elastic behavior, where the elastic properties are given with respect to the crystallographic directions, and (2) plastic deformation governed primarily by crystallographic slip on a limited number of slip planes.2,14

The elastic component of the deformation in the crystalline phase was characterized by a fourth-order anisotropically elastic modulus tensor that linearly related the elastic Green–Lagrangian strain tensor and the elastic second Piola–Kirchhoff stress. The anisotropically elastic properties were coupled to the crystallographic lattice directions in the configuration that resulted from instantaneous unloading of the current configuration.

The viscoplastic behavior of the crystalline phase was described by a rate-dependent crystal plasticity model, where the plastic flow rate of the crystalline lamella, consisting of a single crystal, was composed of the contributions of all physically distinct slip systems. The shear rate of each slip system was assumed to be related to the corresponding shear stress via a viscoplastic power law relation.30,31

Amorphous Phase

The amorphous phase of semicrystalline polymeric material consists of an assembly of disordered macromolecules that are morphologically constrained by the neighboring crystalline lamellae. Plastic deformation in these domains occurs by the thermally activated rotation of segments. At room temperature, the amorphous phase of HDPE that is the material of interest in this study is in the rubbery regime, with a glass-transition temperature \( T_g \) close to \(-70 \, ^\circ\text{C}\). However, the amorphous phase of other semicrystalline polymers such as nylon, with a higher \( T_g \), has to be modeled as elasto–viscoplastic.*

The initial elastic resistance of the rubbery amorphous phase is well below the elastic resistance of the crystalline domain. The elastic deformations of the amorphous domains are modeled by a generalized neo-Hookean relationship, characterized by a shear modulus and a bulk modulus.

A relatively strain rate-insensitive power law relation between an effective shear strain rate and an effective shear stress was used.9 The plastic rate of stretching was defined by an associated flow rule. The Arruda–Boyce eight-chain network model of rubber elasticity32 was used to account for orientation-induced strain hardening.33, 34

Composite Inclusion Model

The mechanical behavior at the mesoscopic level was modeled by an aggregate of layered two-phase composite inclusions, as was proposed by Lee et al.9,10 for rigid/viscoplastic material behavior. Each separate composite inclusion consisted of a crystalline lamellae that was mechanically coupled to its corresponding amorphous layer. The stress and deformation fields within each phase were assumed to be homogeneous piecewise; however, they may differ between the two coupled phases. The inclusion-averaged deformation gradient and the inclusion-averaged Cauchy stress were given by the volume-weighted average of the respective phases. It was assumed that the crystalline and amorphous components remained fully mechanically coupled. Interface compatibility within the composite inclusion and traction equilibrium across the interface were enforced.

To relate the volume-averaged mechanical behavior of each composite inclusion to the imposed boundary conditions for an aggregate of inclusions, a hybrid local–global interaction law was used. This class of hybrid-inclusion models was introduced by Lee et al.9,10 for rigid/viscoplastic composite inclusions.

In the current hybrid interaction model, local–global compatibility conditions were assumed for the projections of the inclusion-averaged fields for which intra-inclusion equilibrium conditions were formulated. These projections were assumed to equal the corresponding projections of an auxiliary deformation-like tensor.15 Inversely, local–global equilibrium conditions were chosen for the components of the inclusion-averaged fields that were subjected to intra-inclusion compatibility.

A more detailed description of the composite inclusion model is presented elsewhere.15 Some aspects of the finite element implementation in ABAQUS35 are given in Van Dommelen et al.36

Deformation Modes

Initially isotropic and anisotropic material was represented by sets of random and preferential...
orientations, respectively. The mechanical response of these aggregates was investigated for various deformation modes. For this purpose, an aggregate of composite inclusions, represented by a set of crystallographic orientations and corresponding lamellar orientations, was subjected to constant strain rate uniaxial tension in the three principal directions $e_i$ of the material coordinate system:

$$\ddot{R} = I; \quad \ddot{U}_{ii} = \lambda(t); \quad i = 1 \lor i = 2 \lor i = 3,$$  \hspace{1cm} (1)

with

$$\lambda(t) = \exp(\dot{\varepsilon}t),$$  \hspace{1cm} (2)

where $\dot{\varepsilon}$ was set equal to the material reference shear rate $\gamma_0$, $\ddot{R}$ was the mesoscopic rotation tensor, and $\ddot{U}$ was the corresponding right stretch tensor. Furthermore, the components of the mesoscopic Cauchy stress tensor $\bar{\sigma}$ should satisfy:

$$\bar{\sigma}_{12} = \bar{\sigma}_{13} = \bar{\sigma}_{23} = 0; \quad j \in \{1,2,3\} \neq i.$$  \hspace{1cm} (3)

In another test case, pure shear deformation was applied by prescribing one of the basic shear components $ij$ of the (symmetric) right stretch tensor:

$$\ddot{R} = I; \quad \ddot{U}_{ij} = \gamma(t); \quad ij = 12 \lor ij = 13 \lor ij = 23,$$  \hspace{1cm} (4)

with

$$\gamma(t) = \frac{1}{2} \sqrt{3} \dot{\gamma}_0 t,$$  \hspace{1cm} (5)

and

$$\bar{\sigma}_{11} = \bar{\sigma}_{22} = \bar{\sigma}_{33} = \bar{\sigma}_{kl} = 0; \quad kl \in \{12,13,23\} \neq ij.$$  \hspace{1cm} (6)

**MODELING RESULTS**

The mechanical responses of various microstructures are investigated for different deformation modes, using the composite inclusion model. In the following discussion, discrete orientation sets were assumed to represent polyethylene with either a spherulitic or a stacked lamellar morphology and with twisted or untwisted lamellae. All properties of the individual crystalline and amorphous phases were assumed to be similar to the properties determined in ref. 15 for initially isotropic HDPE; that is, except for the initial crystallographic and lamellar orientations, all other properties were assumed to be similar for all morphologies. Therefore, all differences in mechanical response between spherulitic and oriented material were entirely the result of differences in the orientation of the microstructural constituents.

**Randomly Oriented Material**

The local spherulitic structure of melt-crystallized HDPE was represented by an aggregate of 125 composite inclusions with randomly generated initial orientations of the crystallographic phases, having an orthorhombic lattice. The distribution of orientations of the principal lattice directions is represented in Figure 6(a–c). Experimental studies of melt-crystallized polyethylene showed that lamellar surfaces are of the $(h0l)$-type, where the angle between the chain direction $c$ and the lamellar normal $n$ vary between $20^\circ$ and $40^\circ$. Gautam et al. found, by molecular simulations, the (201) planes to provide the lowest amorphous/crystalline interface energy. Here, the initial angle between $c_0$ and $n'_0$ was set at $35^\circ$, corresponding to the (201) planes. The resulting initial orientations of the lamellar normals are shown in Figure 6(d). Because the distribution of the crystallographic orientations is random, the mechanical behavior of this aggregate will be quasi-isotropic. The aggregate was subjected to the boundary conditions for different deformation modes, as previously described. The equivalent mesoscopic stress $\bar{\sigma}^m = \frac{1}{2} \bar{\sigma}^{22} \bar{\sigma}^{dd}$, with $\bar{\sigma}^{dd} = \bar{\sigma} + \bar{\rho}t$ the mesoscopic deviatoric stress tensor and $\bar{\rho} = \frac{1}{3} \text{tr}(\bar{\sigma})$ the mesoscopic hydrostatic pressure, as a function of the imposed deformation, is represented in Figure 6(e) and 6(f). The equivalent stresses were normalized by the lowest slip resistance, $\tau_0 = 8$ MPa. The number of inclusions within an aggregate should be sufficiently large to expel the influence of the particular set of initial inclusions and to mimic a truly isotropic material behavior. Here, the number of inclusions that comprise an aggregate was limited to 125, hence a quasi-isotropic response was obtained. The influence of the aggregate size on the mesoscopic behavior was examined in Van Dommelen et al. A geometrically refined multiscale model of spherulitic material was presented in Van Dommelen et al.
Twisted Stacked Lamellae

A stacked lamellar structure of HDPE material was assumed, where all crystallographic [010]-directions, which are the growth directions of the lamellae\(^1\), were arranged perpendicular to a global (fiber) symmetry direction. Furthermore, lamellar twisting, locally around the crystallographic \(b\)-axes, was assumed. The twisted lamellar structure is schematically illustrated in Figure 7. A set of 125 crystallographic orientations was chosen that represents the described structure, with a certain additional random deviation. These orientations are represented by equal area pole figures in Figure 8(a–c), where the view direction is the global 1-direction, which equals the fiber symmetry direction. All (010) poles were chosen to lie randomly within the 23-plane, that is, lie in the outer region of the pole figure. Additionally, lamellar twisting was modeled by a random rotation around the local \(b\)-axis. A similar nature of the chain tilt angle was assumed for the stacked lamellar structure and for the spherulitic polyethylene. Hence, the initial angle between \(c_0\) and \(n_0\) is set at 35°, corresponding to the \{201\} planes. The initial orientations of the lamellar normals are shown in Figure 8(d).

The equivalent mesoscopic stress obtained, as a function of the imposed deformation, when loaded in different modes, is represented in Figure 8(e) and 8(f). The constitutive behavior at the mesoscopic scale should be transversely isotropic. Therefore, all differences in mechanical response in the 22 and 33 tensile directions and the 12 and 13 shear directions were of statistical origin. For this material, with stacked twisted lamellae, only a mild anisotropy was observed. The yield stresses in the transverse 22 and 33 tensile directions and the 23 shear deformation mode were found to be slightly elevated, whereas a small reduction of the strength in the 11 tensile and the 12 and 13 shear deformation modes was observed. Little difference between tensile and compressive deformation was found.

Untwisted Stacked Lamellae

When the melt was subjected to a sufficient amount of shear flow before crystallization, the molecular chains became extended, and a stacked lamellar structure without the twisting nature
was formed,\cite{27,40} as is schematically represented in Figure 9. For this material, the lamellar normals were aligned, preferably parallel to the fiber symmetry direction. The crystallographic $b$-axes were again assumed to lie randomly within the plane that is perpendicular to the axis of symmetry. The chain tilt angle correlates to the twisting nature of the lamellae.\cite{41} In the absence of twisting, the molecular chains were assumed to be more aligned with the lamellar normals, which were again assumed to be of the $(h0l)$-type. Reduced chain tilt angles of $19^\circ$ were assumed. The crystallographic and morphological orientations of the sets of 125 composite inclusions representing these structures, as well as the corresponding mechanical response, are shown in Figure 10. The alignment of molecular chains toward the 1-direction induced increased post-yield hardening when loaded in tension in the 1-direction. For compressive loading in the symmetry direction, post-yield softening was observed. A relatively low modulus for the 12 and 13 shearing modes was observed, corresponding well to experimental observations of the mechanical response of HDPE with a stacked lamellar morphology.\cite{19}

In the absence of sufficient strain hardening at the mesoscopic scale, a macroscopic tensile bar will show necking behavior. For both the randomly oriented material and the aggregate with a twisted stacked lamellar morphology, the mesoscopic analysis indicates that a neck will be formed macroscopically. For the untwisted material, the macroscopic response will remain stable when loaded in the extrusion direction because of the post-yield strain hardening. In the other directions, however, even more severe necking may be expected.

Macroscopic Response

The localization behavior of polyethylene with an untwisted stacked lamellar morphology in tensile tests was simulated with a finite element approach. The three-dimensional finite element mesh of a dumbbell-shaped tensile bar is shown in Figure 11. The width of the test section was five times the sample thickness. The $x$-displacements of the nodes at the right boundary were prescribed at a constant velocity, $\nu = \dot{\varepsilon}L_0$, with $\dot{\varepsilon} = 10^{-3}$ s$^{-1}$. The mesh contained 64 twenty-noded
reduced integration quadratic brick elements. The crystallographic and lamellar orientations in each integration point were generated in a local coordinate system $\mathbf{e}_i$. For extruded material, all local $\mathbf{e}_i$-directions, corresponding to the view direction in the previously discussed pole figures (Fig. 10), were assumed to be parallel to the global extrusion direction that is oriented within the plane of the sample. Different initial angles, $\varphi$, between the extrusion direction and the loading direction of the tensile bar were assumed. In each integration point, an aggregate with orientations similar to the distribution in Figure 10 was generated. Each aggregate consisted of 64 composite inclusions. In Figure 12(a–c), the magnitude of plastic deformation obtained, $\tilde{\varepsilon}_p = \frac{1}{\sqrt{3}} \bar{\varepsilon}_p$, with $\bar{\varepsilon}_p$ the mesoscopic plastic strain tensor, is displayed for $\dot{\varepsilon}_t = 0.45$. The corresponding stress–strain behavior is shown in Figure 12(d) and corresponds reasonably well with the experimentally observed response (Fig. 5(b)). For the tensile bar with $\varphi = 90^\circ$ (i.e., the extrusion direction was perpendicular to the sample long axis), due to the absence of sufficient strain hardening at the mesoscopic level, macroscopic softening was observed, which will lead to the onset of necking. In the localized zone, a strong reduction of the sample thickness was observed. For the sample with the tensile axis aligned with the extrusion direction, the mesoscopic post-yield hardening stabilized the deformation, spreading it out over a larger zone. Therefore, the deformation remained homogeneous and no neck was formed. Significant deformation hardening was observed, also macroscopically, corresponding to the experimentally observed response (Fig. 5(b)). For the sample that was oriented with the loading axis at $45^\circ$ with the extrusion direction, a large reduction of the width of the test section was found. Macroscopically, a reduction of the initial modulus was observed. Additionally, in Figure 13, the width and thickness components of the logarithmic strain tensor are displayed. The localization observed for the $90^\circ$ sample originated from a contraction in the thickness direction. The strains in the molecular chain direction (i.e., the width di-
rection) remained small. In the 90° situation, deformation occurred predominantly in the width direction.

CONCLUSIONS

The deformation of semicrystalline polymeric materials is the result of the interplay of various effects and mechanisms at different levels. A universal prediction of the constitutive behavior of these materials would require a coupled and detailed modeling of the various deformation mechanisms and criteria for the different failure modes, which is at present still not feasible. In this study, the sole influence of the microstructure on the qualitative mechanical response, represented by the orientations of the crystalline lamellae, was investigated. A multiscale numerical modeling framework was used to simulate the effect of the stacked lamellar microstructure as it exists in extruded semicrystalline materials. This stacked lamellar morphology gives rise to a strong influence of the extrusion direction with respect to the loading direction on stability and localization in tensile experiments. The microscopic morphology of extruded HDPE material was characterized by WAXD experiments, which showed a strong alignment of molecular chains with the extrusion direction for specimens produced with a large draw ratio of the melt. The sharp texture of (001) poles led to the conclusion that no lamellar twisting was present.

Orientational input for the composite inclusion model was generated based on the X-ray observations. The effect of lamellar twisting was investigated for aggregates with a fiber symmetry. An elevated yield stress and increased post-yield hardening were observed for untwisted material in the extrusion direction. The material was found to soften under compression after yielding at an elevated stress. These effects were, however, not present for the aggregate with a twisted
microstructure, for which the molecular chains were not aligned with the loading direction.

A macroscopic tensile bar was modeled by finite elements. The microstructure-induced deformation hardening in the extrusion direction was found to stabilize the macrostructure when loaded in the flow direction. If the extrusion direction was perpendicular to the long axis of the macroscopic tensile bar, the onset of necking was observed, accompanied by a strong reduction of the specimen thickness. When loaded at 45°, the tensile deformation was accompanied by a large reduction of the specimen width. Thus, the microstructure-based model can explain the experimentally observed orientation-dependence of extruded semicrystalline material, as was demonstrated in Figure 1.

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REFERENCES AND NOTES