Micromechanical Modeling of Anisotropic Behavior of Oriented Semicrystalline Polymers

Mohsen Mirkhalaf,1 Johannes A. W. van Dommelen,1 Leon E. Govaert,1 Jevan Furmanski,2 Marc G. D. Geers1
1Department of Mechanical Engineering, Eindhoven University of Technology, P.O. Box 513, 5600 MB Eindhoven, The Netherlands
2ExxonMobil Research and Engineering, Annandale, New Jersey, 08801, United States
Correspondence to: J. A. W. van Dommelen (E-mail: j.a.w.v.dommelen@tue.nl)

Received 15 October 2018; accepted 23 January 2019
DOI: 10.1002/polb.24791

ABSTRACT: Some manufacturing processes of polymeric materials, such as injection molding or film blowing, cause the final product to be highly anisotropic. In this study, the mechanical behavior of drawn polyethylene (PE) tapes is investigated via micromechanical modeling. An elasto-viscoplastic micromechanical model, developed within the framework of the so-called composite inclusion model, is presented to capture the anisotropic behavior of oriented semicrystalline PE. Two different phases, namely amorphous and crystalline (both described by elasto-viscoplastic constitutive models), are considered at the microstructural level. The initial oriented crystallographic structure of the drawn tapes is taken into account. It was previously shown by Sedighiamiri et al. (Comp. Mater. Sci. 2014, 82, 415) that by only considering the oriented crystallographic structure, it is not possible to capture the macroscopic anisotropic behavior of drawn tapes. The main contribution of this study is the development of an anisotropic model for the amorphous phase within the micromechanical framework. An Eindhoven glassy polymer (EGP)-based model including different sources of anisotropy, namely anisotropic elasticity, internal stress in the elastic network and anisotropic viscoplasticity, is developed for the amorphous phase and incorporated into the micromechanical model. Comparisons against experimental results reveal remarkable improvements of the model predictions (compared to micromechanical model predictions including isotropic amorphous domains) and thus the significance of the amorphous phase anisotropy on the overall behavior of drawn PE tapes. © 2019 The Authors.

KEYWORDS: anisotropic amorphous phase; micromechanical model; oriented polyethylene; semicrystalline polymers; structure-property relations microstructure orientation

INTRODUCTION Polymers have become an important and highly applicable type of materials due to their interesting chemical, optical, thermal, and mechanical properties. Some polymeric materials develop a semicrystalline microstructure during the cooling process from the molten state. The microstructure of semicrystalline polymers is composed of different phases, a crystalline phase and an amorphous phase.1

During some manufacturing processes for polymeric products, such as injection molding or film blowing, the material is under shear and elongational flow resulting in an oriented microstructure after crystallization, which leads to anisotropic mechanical behavior. Semicrystalline polymers are used in different industries and in load-bearing applications. Load and lifetime assessments require to predict the orientation dependent thermomechanical behavior of these materials in a quantitative manner, based on the inherently hierarchical heterogeneous structure of semicrystalline polymers, using multiscale simulations.

The history of micromechanical modeling of semicrystalline polymers dates back to the 1990’s. Parks and Ahzi2 and Lee et al.3,4 were the pioneers in modeling the mechanical behavior of semicrystalline polymers within a micromechanical framework. Parks and Ahzi2 developed a fully crystalline micromechanical rigid-viscoplastic model (named constrained hybrid [CH] model) to describe the deformation behavior and texture evolution of polycrystalline materials. Ahzi et al.5 described large plastic deformation and also texture evolution of different semicrystalline polymers with the CH model proposed by Parks and Ahzi and the reformulated version of the CH model by Lee et al.,6 in which idealized fully crystalline materials were considered and the deformation mechanism was assumed to be crystallographic slip only. A rigid
The long-term and short-term mechanical behavior of anisotropic PET. For that purpose, an internal stress was incorporated into the constitutive description of the amorphous phase. The same approach is followed in this study. However, the anisotropic model developed in this study has an additional anisotropic viscoplastic flow rule which is developed using the anisotropic effective stress of Hill.15,16 The effect of the amorphous phase anisotropy is significant and comparisons between the model predictions and experimentally obtained results show significant improvements.

The next section gives the experimental results for oriented PE tape under different loading angles and different strain rates, reported by Sedighiamiri et al.11 In this section, the micromechanical framework, including the original isotropic model for the amorphous phase, is explained. Based on the results with this model and comparisons to experimental results, the motivation for extending the constitutive model for the amorphous phase is discussed. Section explains various sources of anisotropy in oriented PE tapes and describes the anisotropic model for the amorphous phase, including an internal stress into the elastic network, an anisotropic viscoplastic flow rule and anisotropic elasticity. The material properties for the individual microstructural phases are given and the Hill parameters and prestretch factor for the amorphous phase are quantified. Comparisons between the simulations and experimental results are made. Section summarizes the conclusions drawn from this study.

EXPERIMENTAL RESULTS

In this study, the mechanical anisotropy of hot-drawn PE tapes, in which the crystalline domains show a preferred orientation distribution, is modeled. Sedighiamiri et al.11 conducted tensile experiments on HDPE tape with a draw ratio of 4 (\( \lambda = 4 \)) at different loading angles (\( \phi = 0^\circ, 20^\circ, 50^\circ \)) and under different strain rates. WAXS experiments were conducted and the weight fraction of the crystalline phase of the oriented Polyethylene tape was determined to be 66%. Additional WAXS experiments were conducted to obtain the crystallographic orientation distribution. Uniaxial tensile tests were performed on dog-bone-shaped samples that were cut from the tapes at different angles to investigate the orientation effects. Three different strain rates, within the quasi-static range, were applied (\( \dot{\varepsilon} = 0.0001 \text{ s}^{-1}, 0.001 \text{ s}^{-1}, 0.001 \text{ s}^{-1} \)). Figure 1 shows a schematic representation of the samples cut from the oriented tapes at different angles with respect to the machine direction, together with the experimentally obtained stress–strain curves at room temperature.

A strong degree of dependence on both loading angle and strain rate is observed in the stress–strain curves.

MICROMECHANICAL MODEL

In this section, the micromechanical model, used in this study, including the original isotropic model for the amorphous
phase, is presented.\textsuperscript{10} The model has four key components that constitute the basis of the model:

- two-phase composite inclusion,
- inclusion interaction law,
- constitutive model of the crystalline phase, and
- constitutive model of the amorphous phase.

Each of the aforementioned aspects of the micromechanical model will be explained in detail in the following.

**Two-Phase Composite Inclusion**

Lee et al.\textsuperscript{3,4} proposed a framework for modeling the rigid/visco-plastic mechanical behavior of semicrystalline materials. This framework uses a layered two-phase composite inclusion as the basis of the model, whereby an aggregate of inclusions constitutes the whole material. This approach is used in this study. Each crystalline lamella and the corresponding attached amorphous layer, as shown in Figure 2(a), is considered as an inclusion.

The volume-averaging relations for the deformation gradient tensor and Cauchy stress tensor apply:

\[ F_I = f_a F_a + (1 - f_a) F_c, \]  
\[ \sigma_I = f_a \sigma_a + (1 - f_a) \sigma_c, \]  

where \( F \) represents the corresponding volume change ratio (det(\( F \))). The superscripts \( a, c \) and \( l \) refer to the amorphous phase, the crystalline phase and the inclusion, respectively. Compatibility of deformation gradients and traction equilibrium are considered at the interface between the constituents. If \( \mathbf{n} \) is the unit normal vector on the interface in the deformed configuration, traction equilibrium is obtained by:

\[ \sigma_a \mathbf{n} = \sigma_c \mathbf{n} = \sigma_I \mathbf{n}, \]  

The compatibility condition is obtained by:

\[ F_a \mathbf{x}_0 = F_c \mathbf{x}_0 = F_I \mathbf{x}_0, \]  

where \( \mathbf{x}_0 \) denotes an arbitrary vector in the interface plane in the reference configuration.

**Inclusion Interaction Law**

The semicrystalline polymer is considered as an aggregate of randomly or preferentially oriented inclusions for isotropic and anisotropic materials, respectively. Figure 2(b) depicts a schematic representation of an aggregate of uniformly distributed inclusions. A local–global interaction law is defined to relate the mechanical behavior of each composite inclusion to the imposed boundary conditions of the aggregate of inclusions. In this work, a hybrid interaction model, which is basically an intermediate approach between the Taylor\textsuperscript{33} and Sachs\textsuperscript{34} interaction laws and referred to as the \( \mathbf{U} \)-inclusion model, is used.\textsuperscript{1} In the hybrid interaction law, six auxiliary...
deformation-like unknowns, $U$, are introduced for which the following local–global relation holds:

$$U_i^{(l)} = U_0, \quad i = 1, \ldots, N^l$$

where $N^l$ is the number of inclusions and $U$ is the right stretch tensor, such that $F = R \cdot U$, where $R$ is the rotation tensor. Using the global (macroscopic) Cauchy stress tensor $\sigma$, and the global rotation tensor, $R$, the following local–global relations hold:

$$\sigma_i^{(l)} \cdot x_i^{(l)} = \sigma_i \cdot x_i, \quad i = 1, \ldots, N^l, \quad (7)$$

$$R_i^{(l)} = R, \quad i = 1, \ldots, N^l, \quad (8)$$

in combination with the following consistency conditions:

$$\sigma = \frac{1}{J} \sum_{i=1}^{N^l} f_i^{(l)} \sigma_i^{(l)}, \quad (9)$$

$$U = \left( \frac{1}{J} \sum_{i=1}^{N^l} f_i^{(l)} U_i^{(l)} \right)^{1/3}, \quad (10)$$

where $J$ and $J_\Sigma$ are given by:

$$J = \sum_{i=1}^{N^l} f_i^{(l)} f_i^{(l)}, \quad (11)$$

$$J_\Sigma = \det \left( \sum_{i=1}^{N^l} f_i^{(l)} F_i^{(l)} \right) \Sigma, \quad (12)$$

Before detailing the constitutive relations for the crystalline and amorphous phases, it is necessary to mention that for the both phases, the deformation gradient is assumed to be multiplicatively decomposed into its elastic and plastic components, depicted by subscripts $e$ and $p$, respectively:

$$F = F_e \cdot F_p. \quad (13)$$

This decomposition suggests a local unstressed intermediate configuration obtained by elastic unloading from the final deformed configuration. The velocity gradient is given by:

$$L = L_e + F_e \cdot F_p \cdot F^{-1}, \quad (14)$$

where the elastic velocity gradient, $L_e$, and plastic velocity gradient, $L_p$, are defined by:

$$L_e = F_e \cdot F^{-1}, \quad (15)$$

$$L_p = F_p \cdot F^{-1}. \quad (16)$$

C}rystalline Phase Constitutive Model

The crystalline domain of HDPE has an orthorhombic structure with anisotropic properties, including a high elastic modulus in the chain direction. The elastic stress–strain relation for the crystalline phase is given by:

$$S_e = 4 C_e : E_c, \quad (17)$$

where the elastic second Piola-Kirchhoff stress, $S_e$, is defined by:

$$S_e^e = J_e^e \cdot F_e^{-1} \cdot \sigma^e \cdot F_e^{-T} \cdot C_e. \quad (18)$$

Crystalline lamella

Amorphous phase

(a)

FIGURE 2 (a): Composite inclusion consisting of a crystalline lamella and an amorphous layer. (b) Schematic representation of an aggregate of randomly distributed composite inclusions. [Color figure can be viewed at wileyonlinelibrary.com]
\[
E^* = \frac{1}{2} \left( F^T \varepsilon - F \right).
\]  
(19)

where \( I \) is the second-order identity tensor. In eq 17, \( ^C \varepsilon \) denotes the elasticity tensor which, in matrix form using the Voigt notation (where the stress components are ordered as \( \sigma^T = [\sigma_{11} \quad \sigma_{22} \quad \sigma_{33} \quad \sigma_{12} \quad \sigma_{13} \quad \sigma_{23}] \)), can be represented by:

\[
C = \begin{bmatrix}
    c_{11} & c_{12} & c_{13} & 0 & 0 & 0 \\
    c_{12} & c_{22} & c_{23} & 0 & 0 & 0 \\
    c_{13} & c_{23} & c_{33} & 0 & 0 & 0 \\
    0 & 0 & 0 & c_{44} & 0 & 0 \\
    0 & 0 & 0 & 0 & c_{55} & 0 \\
    0 & 0 & 0 & 0 & 0 & c_{66}
\end{bmatrix}.
\]  
(20)

The plastic behavior of the crystalline phase is assumed to occur by crystallographic slip up to moderate strains.\textsuperscript{18–21} In the micromechanical model, the slip mode is considered to be merely fine slip. A rate-dependent crystal plasticity model is used to describe the viscoplastic part of the deformation in the crystalline lamellae. The contribution of all physically distinct slip systems is taken into account:

\[
L^p_{i} = F^p_{i} [F^p_{i}]^{-1} = \sum_{\alpha=1}^{N_s} \gamma^\alpha P^\alpha_0,
\]  
(21)

where \( N_s \) is the number of physically distinct slip systems, \( P^\alpha_0 \) is the nonsymmetric Schmid tensor given by:

\[
P^\alpha_0 = s^\alpha \otimes n^\alpha_0,
\]  
(22)

where \( s^\alpha \) is the unit slip direction and \( n^\alpha_0 \) is the unit slip plane normal of slip system \( \alpha \), both in the reference configuration. In eq 21, \( \gamma^\alpha \) is the shear rate of slip system \( \alpha \). When a single relaxation process is considered, \( \gamma^\alpha \) is defined using an Eyring flow rule:\textsuperscript{10}

\[
\dot{\gamma}^\alpha = \dot{\gamma}^\alpha_0 \exp \left( \frac{\Delta U^\alpha_{\text{act}}}{RT} \right) \sinh \left( \frac{\tau^0_\alpha}{\tau^\alpha_0} \right) \exp \left( \frac{\mu^\alpha \sigma^\alpha_n}{\tau^\alpha_0} \right),
\]  
(23)

where \( \dot{\gamma}^\alpha_0 \) is the reference shear rate, \( \tau^0_\alpha \) is the reference shear strength, \( \tau^\alpha_0 \) is the resolved shear stress. Part (a) of eq 23 describes the temperature dependence of the slip kinetics where, \( \Delta U^\alpha_{\text{act}} \) is the activation energy of each slip system, \( R \) is the universal gas constant and \( T \) is the absolute temperature. The stress dependence of the plastic flow is captured in part (b) of the equation. Finally, part (c) captures the effect of the normal stress of each slip plane \( (\sigma^\alpha_n) \) on the plastic flow, where the parameter \( \mu^\alpha \) is the coefficient of normal stress dependency. Experimental results of tensile yield kinetics for initially isotropic Polyethylene at different temperatures and strain rates show that two relaxation processes contribute to the viscoplastic flow behavior.\textsuperscript{10} Based on these observations, two different flow processes are distinguished, as shown in Figure 3.

The extended slip rate considering both processes I and II is given by:\textsuperscript{10}

\[
\dot{\gamma}^\alpha = \dot{\gamma}^\alpha_0 \exp \left( \frac{\Delta U^\alpha_{\text{act}}}{RT} \right) \sinh \left( \frac{\tau^0_\alpha}{\tau^\alpha_0} \right) \exp \left( \frac{\mu^\alpha \sigma^\alpha_n}{\tau^\alpha_0} \right),
\]  
(24)

where \( \tau^0_{\alpha} \) represents the shear stress dependence when both processes I and II are activated:

\[
\tau^0_{\alpha} = \tau^0_{\alpha} + \tau^0_{\alpha} \tau^0_{\alpha}.
\]  
(25)

Also, the temperature dependent pre-exponential initial shear rate, \( \dot{\gamma}^\alpha_0_{\alpha} \), includes contributions of two processes:

\[
\dot{\gamma}^\alpha_0_{\alpha} = 2 \exp \left( -\frac{\Delta U^\alpha_{\text{act}}}{RT} \right) \sinh \left( \frac{\tau^0_\alpha}{\tau^\alpha_0} \right) \exp \left( \frac{\mu^\alpha \sigma^\alpha_n}{\tau^\alpha_0} \right).
\]  
(26)

It should be noted that super script ("*) which refers to each slip system is omitted in relations 24–26 to avoid confusions.

**Amorphous Phase Constitutive Model**

In this section, the initial *isotropic* model for the amorphous phase is described. The model is developed based on the Eindhoven glassy polymer (EGP) model\textsuperscript{22,23} for amorphous polymers. A single mode model with two relaxation processes (\( \alpha \) and \( \beta \) which will be referred to I and II, respectively) is adopted. The mechanical analogue of the single mode EGP model is shown in Figure 4.

The total Cauchy stress is additively composed of the driving stress (\( \sigma^\alpha \)) and the hardening stress (\( \sigma^\beta \)):
FIGURE 4 Mechanical analogue of the single mode EGP model. [Color figure can be viewed at wileyonlinelibrary.com]

\[ \sigma^p = \sigma_s^p + \sigma_a^p. \]  

(27)

The driving stress is dependent on the elastic deformation gradient and is composed of hydrostatic and deviatoric contributions indicated by superscript \( h \) and \( d \), respectively, given by:

\[ \sigma_{s}^{h} = - p a I; \quad p a = - K^a (f_c^a - 1), \]  

(28)

\[ \sigma_{a}^{d} = G^a B_e. \]  

(29)

where \( K^a \) is the bulk modulus and \( G^a \) is the shear modulus. The isochoric elastic left Cauchy-Green deformation tensor is defined by:

\[ B_e = \int_e \frac{2}{3} F_e F_e^{T}. \]  

(30)

The hardening stress is dependent on the total isochoric deformation gradient:

\[ \sigma^p = G_c B_e. \]  

(31)

where \( G_c \) is the hardening modulus. A constitutive relation for the plastic deformation rate tensor is introduced in order to fully describe the deformation behavior of the amorphous phase. The plastic deformation rate tensor is defined by:

\[ D_p = \text{sym}(F_e L_p F_e^{-1}). \]  

(32)

In this model, the plastic stretching tensor is given by:

\[ D_p = \frac{\sigma_{a}^{d}}{2\eta}, \]  

(33)

where \( \eta \) is the viscosity of the amorphous phase given by:

\[ \eta = \frac{\tau}{\mu^a}. \]  

(34)

and \( \tau \) is the effective stress given by:

\[ \tau = \sqrt{1 \over 2} \left( \sigma_a^d : \sigma_a^d \right). \]  

(35)

In eq 34, \( \dot{\gamma}_p \) is the plastic shear rate which for two relaxation processes is given by:

\[
\dot{\gamma}_p = \left( \frac{\dot{\gamma}_0^{a,i} \sinh \left( \frac{\tau}{\dot{\gamma}_0^{a,i}} \right) \dot{\gamma}_0^{a,i+II} \sinh \left( \frac{\tau}{\dot{\gamma}_0^{a,i+II}} \right)} {\dot{\gamma}_0^{a,i} \sinh \left( \frac{\tau}{\dot{\gamma}_0^{a,i}} \right) + \dot{\gamma}_0^{a,i+II} \sinh \left( \frac{\tau}{\dot{\gamma}_0^{a,i+II}} \right)} \right) \exp \left( -\frac{\mu \eta^a}{\dot{\gamma}_0^a} \right), \tag{36}
\]

where

\[
\dot{\gamma}_0^{a,i+II} = 2 \exp \left( -\frac{\tau_0^{a,i+II} / \dot{\gamma}_0^{a,i+II}} {\tau_0^{a,i+II} / \dot{\gamma}_0^{a,i+II}} \right), \tag{37}
\]

\[
\tau_0^{a,i+II} = \tau_0^a + \tau_0^i. \tag{38}
\]

The superscripts \( I \) and \( II \) denote two relaxation processes, \( \tau \) is the effective stress, \( \mu \) is the pressure dependency coefficient, and \( \tau_0^i \) is the shear strength of the amorphous phase. Kanters et al.\(^{25}\) compared this approach with one that involves separate kinematics for the two relaxation processes and found that there is only a small difference between the aforementioned approaches.

Initial Results

The mechanical behavior of oriented tape is first predicted with the micro-mechanical model described in the previous section using an isotropic elasto-viscoplastic constitutive model for the amorphous phase. Based on wide angle X-ray scattering measurements, the initial orientation distribution was generated. The pole figures of the crystallographic and morphological orientations were generated to be consistent with the experimentally obtained pole figures.\(^{11}\) As a quantitative measure for the orientation, the Hermans orientation factor was used:\(^{25}\)

\[
f_H = \frac{3 \cos^2 \Phi - 1}{2}, \tag{39}
\]

where \( \Phi \) represents the angle between the crystallographic axis and a reference axis. The Hermans orientation factor varies between \(-0.5 \) and \(1\), where \( f_H = -0.5 \) represents an orthogonal orientation, \( f_H = 0 \) is obtained for a random orientation and \( f_H = 1 \) indicates a perfectly uniaxial orientation. Hermans factors are used to check if the generated samples properly represent the material. Figure 5 depicts pole figures of the crystallographic and lamellar orientations of oriented polyethylene tape with a draw ratio factor of \(4 (\lambda = 4)\).

Table 1 reveals that the Hermans factors of the created samples are close to the experimental ones. In the study by Sedighiamiri, et al.,\(^{11}\) the volume fraction of the crystalline phase was estimated to be 60%. Before starting the simulations, the material properties of the crystalline lamellae and the amorphous domain are first presented.

Crystalline Phase Properties

The crystalline domains of PE possess an orthorhombic structure. Table 2 gives the lattice parameters of the crystalline lamellae.\(^{26,27}\)
The elastic properties of the crystalline phase were obtained from Tashiro et al.28 and are given in Table 3. The yield kinetics of the slip systems were determined such that the experimentally obtained response of both the oriented and the original isotropic material10 could be described for a range of strain rates. Here, the slip kinetics of the various slip systems were chosen to be parallel. Table 4 gives the values of slip kinetics used in this study.

**Amorphous Phase Properties**

The isotropic elastic material properties of the amorphous phase are defined as $G^\prime = 65$ MPa and $K^\prime = 3000$ MPa.10 The viscoplastic (Eyring) properties for the amorphous domain are given in Table 5.10 Figure 6 shows the total (sum of $\alpha$ and $\beta$ relaxation processes) yield kinetics of the slip systems and amorphous phase at room temperature.

A comparison between the experimental results and model predictions obtained with the oriented crystallographic structure and an isotropic amorphous phase is given in Figure 7 for the three loading angles and three strain rates. All the model predictions in this study are converted from true stress–strain to engineering stress–strain curves to make a direct comparison to the experimental results.

The discrepancy between the experimental results and simulations increases by increasing the strain rate and decreasing the loading angle. It should also be emphasized that in the simulations conducted in this study, the generated crystallographic orientation distribution well represents the crystallographic texture considering the WAXS measurements. However, as opposed to isotropic polymeric materials which have a higher compressive yield stress than their tensile yield stress, oriented polymers may show a considerable higher tensile yield strength in comparison to their compressive yield strength.29,30 The yield behavior and Bauschinger effect in injection molded PE at different loading angles and under different strain rates were studied by Senden et al.30 Senden et al.31 modeled a strong Bauschinger effect in oriented polycarbonate using an EGP-based model including a viscous hardening branch. Consequently and considering the aforementioned observed Bauschinger effect in oriented PE, it was concluded that the amorphous phase orientation is not negligible and should be taken into account.

**ANISOTROPIC MODEL FOR THE AMORPHOUS PHASE**

There are different sources of anisotropy in oriented PE:

- Anisotropy associated with the crystalline lamellae:
  - Anisotropic elastic properties of each crystalline lamella;
  - Anisotropic viscoplastic properties of each crystalline lamella due to crystallographic slip processes;
- Anisotropy associated with the amorphous phase:
  - Anisotropic elasticity;
  - Internal stress due to the oriented elastic network;
  - Anisotropic viscoplastic flow;
- Difference in properties between oriented crystalline lamellae and amorphous domains.

In this study, an anisotropic model is developed for the amorphous phases in the micromechanical model. The three sources of anisotropy are incorporated in the model. In the following, each of these aspects is elaborated in detail.

**Internal Stress in the Elastic Network**

Polymeric materials normally show a pressure-dependent behavior. For isotropic polymers, the yield stress in compression is slightly higher than under tensile conditions.29 On the
contrary, for oriented polymers, the yield stress in tensile loading may be considerably higher than in compression.\textsuperscript{29,30} Senden et al.\textsuperscript{30} studied the yield behavior of injection molded PE and showed a considerably higher yield strength in tension compared to compressive yield strength. Also, it was shown by Senden et al.\textsuperscript{29,31} that by increasing the level of pre-stretch, the difference between tensile and compressive yield stress increases. The internal stress was found to be the main reason for the observed orientation-dependent thermal shrinkage of PET films.\textsuperscript{12} Furthermore, Poluektov et al.\textsuperscript{12} showed that for oriented PET, it is necessary to include an internal stress into the elastic network of the amorphous phase in order to obtain a realistic macroscopic response for the creep compliance. Based on the aforementioned observations, an internal stress is incorporated into the elastic network for the amorphous phase of oriented PE. The internal stress, although different from the classical kinematic hardening, will act as kinematic hardening. In order to include an internal prestress in the elastic elements (describing the molecular network) of the model, a prestretch deformation ($F_{\text{pre}}$) is applied to the hardening branch. Figure 8 depicts schematically how the prestretch deformation gradient is applied to the elastic network branch.

The pre-stretch deformation gradient (i.e. affecting the initial state) in the elastic spring in the driving branch of the model is obtained by equilibrating the amorphous phase. Once the elastic prestretch in the driving branch is obtained, the initial plastic state is obtained so that the total deformation gradient of the driving branch is equal to identity (see Fig. 8). When the material is under deformation ($F^P$), the deformation gradient tensor $\textbf{F}^d$ of the hardening branch is obtained by:

$$\textbf{F}^d = \textbf{F}^P \cdot F_{\text{pre}}^{-1},$$

where $\textbf{F}^d$ is the deformation gradient of the amorphous phase and $F_{\text{pre}}^{-1}$ is the imposed pre-stretching deformation gradient.

### Anisotropic Viscoelastic Flow

As mentioned earlier, strong tension-compression asymmetry is observed in the deformation behavior of oriented polymers.\textsuperscript{29,30} Senden et al.\textsuperscript{31} modeled the mechanical behavior of polycarbonate using an EGP-based model including a viscous strain hardening contribution. Also, the cyclic uniaxial deformation of PC was modeled and a strong Bauschinger effect was properly captured. This indicates the presence of anisotropy in the visco-plastic flow. Besides, in preliminary simulations using the anisotropic constitutive model for the amorphous phase, combining with an internal stress, it was discovered that the viscoplastic flow rule should also be modified from isotropic to anisotropic so that the anisotropic rate dependence of the oriented material is also accurately modeled. The constitutive relation for the plastic deformation rate tensor is written as:

![Figure 6: Total slip kinetics of the crystalline lamellae and the yield kinetics of the amorphous phase (sum of $\alpha$ and $\beta$ processes) at room temperature adopted in the initial simulations. [Color figure can be viewed at wileyonlinelibrary.com]](image)
where $\dot{\gamma}^p$ is the plastic shear rate given by eq 36. For the isotropic flow rule, the effective stress was given by eq 35. In order to modify the flow rule of the model to anisotropic flow, the equivalent stress ($\tau$) is replaced with Hill’s anisotropic effective stress\textsuperscript{15,16} which is given by:\textsuperscript{30,32}

$$\tau^2 = F\left(\sigma_{22}^* - \sigma_{33}^*\right)^2 + G\left(\sigma_{33}^* - \sigma_{11}^*\right)^2 + H\left(\sigma_{11}^* - \sigma_{22}^*\right)^2$$

**TABLE 6** Anisotropic Elastic Properties of the Amorphous Phase of Oriented PE

<table>
<thead>
<tr>
<th>$E_{11}$, $E_{33}$ (MPa)</th>
<th>$\nu_{13}$, $\nu_{23}$</th>
<th>$\nu_{12}$, $\nu_{21}$</th>
<th>$G_{12}$, $G_{13}$, $G_{23}$ (MPa)</th>
</tr>
</thead>
<tbody>
<tr>
<td>200</td>
<td>600</td>
<td>0.1</td>
<td>0.3</td>
</tr>
</tbody>
</table>
where \( \sigma^* \) is the driving stress tensor in the local coordinate system:

\[
\sigma^* = R^* \sigma^0 R^a,
\]

where \( R^* \) is the rotation tensor. The anisotropy parameters in eq 42 are defined by:

\[
\begin{align*}
F &= \frac{1}{2} \left( \frac{1}{R_{11}^2} + \frac{1}{R_{22}^2} - \frac{1}{R_{33}^2} \right), \\
L &= \frac{3}{2} R_{23}^2, \\
G &= \frac{1}{2} \left( \frac{1}{R_{11}^2} + \frac{1}{R_{22}^2} - \frac{1}{R_{33}^2} \right), \\
M &= \frac{3}{2} R_{13}^2, \\
H &= \frac{1}{2} \left( \frac{1}{R_{11}^2} + \frac{1}{R_{22}^2} - \frac{1}{R_{33}^2} \right), \\
N &= \frac{3}{2} R_{12}^2.
\end{align*}
\]

The \( R_{ij} \) factors in relation 44 are the ratios between the corresponding anisotropic yield stress in \( ij \) direction to the reference yield stress. Introducing all \( R_{ij} \) parameters to unity reduces the anisotropic flow rule to the isotropic case with \( \sigma_{ii}^0 = \sqrt{3} \tau \). Accordingly, the plastic shear rate, given in eq 36, is updated as:

\[
\dot{\gamma}^p = \left( \frac{\gamma_{0}}{\gamma_{00}} \sinh \left( \frac{\tau}{\sqrt{3} \sigma_{ii}^0} \right) \right) \exp \left( -\mu p^0 \frac{\gamma_{0}}{\gamma_{00}} \right).
\]

In eq 41, \( \mathbf{N} \) indicates the tensorial direction of flow and is given by:

\[
\mathbf{N} = R^* \mathbf{N}^* R^a,
\]

where \( \mathbf{N}^* \) is given by:

\[
\mathbf{N}^* = \frac{1}{\sqrt{3} \sigma^0} \partial \sigma^0.
\]

The components of this tensor, \( \mathbf{N}^* \), are obtained as:

\[
\begin{align*}
N_{11}^* &= \frac{1}{\sigma_{ii}^0 (G + H) - (G \sigma_{33}^* + H \sigma_{22}^*)}, \\
N_{22}^* &= \frac{1}{\sigma_{ii}^0 (G + H) - (G \sigma_{33}^* + H \sigma_{11}^*)}, \\
N_{33}^* &= \frac{1}{\sigma_{ii}^0 (G + H) - (G \sigma_{22}^* + H \sigma_{11}^*)}, \\
N_{12}^* &= \frac{2}{\sigma_{ii}^0 (N \sigma_{12}^*)}, \\
N_{13}^* &= \frac{2}{\sigma_{ii}^0 (N \sigma_{13}^*)}, \\
N_{23}^* &= \frac{2}{\sigma_{ii}^0 (N \sigma_{23}^*)}.
\end{align*}
\]

**Anisotropic Elasticity**

Due to drawing, the polymer chains in the amorphous phase become preferentially oriented in the draw direction of the tape and as a result, a higher stiffness of the amorphous phase in the draw direction of the tape can be expected. Hence, the elastic isotropic behavior of the amorphous phase needs to be modified to the anisotropic case as well.

The elastic stress–strain relation in the amorphous phase is given by:

**TABLE 8** Hill Anisotropy Parameters and Prestretch Factor for the Oriented Tape (\( \lambda = 4 \))

<table>
<thead>
<tr>
<th>Parameter</th>
<th>( R_{11} )</th>
<th>( R_{22} )</th>
<th>( R_{33} )</th>
<th>( R_{12} )</th>
<th>( R_{13} )</th>
<th>( R_{23} )</th>
<th>( \lambda_0 )</th>
</tr>
</thead>
<tbody>
<tr>
<td>Value</td>
<td>1</td>
<td>1</td>
<td>1.7</td>
<td>1</td>
<td>0.35</td>
<td>0.35</td>
<td>1.2</td>
</tr>
</tbody>
</table>
where $S^a$ is the elastic second Piola-Kirchhoff stress in the driving branch. The driving Cauchy stress is then given by:

$$\sigma^a_s = \frac{1}{Ja} F^{ae} \cdot \sigma_s C^1 F^{ae} T,$$

(50)

with $F^{ae}$ the elastic deformation gradient tensor in the amorphous phase. In eq 49, the elastic Green-Lagrange strain, $E^{ae}$, is defined by:

$$E^{ae} = \frac{1}{2} \left( F^{ae T} - I \right),$$

(51)

where $I$ is the second order identity tensor. In eq 49, $^4C^a$ denotes the stiffness tensor of the amorphous phase which in matrix form, using the Voigt notation, has a similar matrix representation as the stiffness of the crystalline lamellae (eq 20). For an orthotropic material, the compliance matrix, which is the inverse of the stiffness matrix, is given by:

$$S^a = \begin{bmatrix}
\frac{1}{E_{11}} - \frac{\nu_{12}}{E_{12}} & \frac{\nu_{13}}{E_{13}} & 0 & 0 & 0 \\
\frac{\nu_{12}}{E_{12}} & \frac{1}{E_{22}} - \frac{\nu_{23}}{E_{23}} & 0 & 0 & 0 \\
\frac{\nu_{13}}{E_{13}} & \frac{\nu_{23}}{E_{23}} & \frac{1}{E_{33}} & 0 & 0 \\
0 & 0 & 0 & \frac{1}{G_{12}} & 0 \\
0 & 0 & 0 & 0 & \frac{1}{G_{13}} \end{bmatrix},$$

(52)
In order to recover the elastic properties for the amorphous phase, the isotropic case will be perturbed starting from the

In this section, the material properties for the crystalline and amorphous phases of oriented PE are determined. Simulations are conducted using the the extended micro-mechanical model (including the anisotropic model for the amorphous phase) and the results are confronted with experimentally obtained stress-strain curves. Significant improvements in the model predictions are obtained and the comparisons with experimental results show the quantitative impact of the amorphous phase anisotropy.

Material Properties
In this section, the material properties including the elastic properties and yield kinetics of the amorphous phase are determined. The lattice parameters, elastic properties, and the yield kinetics of the crystalline lamellae were given in this section (see Tables 2–4).

Amorphous Phase Properties
In order to recover the elastic properties for the amorphous phase, the isotropic case will be perturbed starting from the Young’s modulus and Poisson’s ratio, extended from the shear modulus and bulk modulus of the amorphous phase in isotropic PE ($G = 65$ MPa and $K = 3000$ MPa)\(^\text{19}\). The Young’s modulus in the draw direction of the tape is gradually increased in order to take into account the effect of drawing on the stiffness of the amorphous phase in the draw direction. It was assumed that the main effect of the drawing process on the stiffness of the amorphous phase appears in the actual draw direction and thus the other two stiffness properties ($E_{11}$, $E_{22}$) are considered almost equal to the stiffness of the un-orientied amorphous phase (in isotropic PE). The results and comparison to experiments showed the assumption (regarding effects of drawing on the stiffness) to be reasonable. The Poisson’s ratios are determined to best capture the experimental results and taking into account the symmetry of the compliance matrix. The shear moduli are also assumed to be the same as the shear modulus of isotropic PE. The elastic material properties of the amorphous phase are given in Table 6 and the viscoplastic (Eyring) properties for the amorphous domain are given in Table 7. The yield kinetics of the amorphous phase can have a large effect on the rate sensitivity of the semicrystalline polymer. The parameter $\tau_0$ of the amorphous yield kinetics was increased to increase the rate dependency of the oriented amorphous phase and as a result, the rate dependency of the drawn PE. The various parameters of the oriented amorphous phase were determined in order to describe the entire stress-strain curve, including the post-yield response for all strain rates and loading angles. The yield kinetics of the amorphous phase and different slip systems of the crystalline lamellae at room temperature are shown in Figure 9.

It should be emphasized that for determination of elastic and viscoplastic properties of the amorphous phase, the experimental results at strain rates of $\dot{\varepsilon} = 10^{-4}$ s$^{-1}$ and $\dot{\varepsilon} = 10^{-3}$ s$^{-1}$ are used.

Simulations and Comparison to Experiments
In order to understand the effect of the Hill parameters and the prestretch factor on the deformation behavior of the oriented tape at different loading angles, parameter sensitivity analyses were performed and the effects of the mentioned parameters were investigated. According to the sensitivity analyses, the effect of the Hill parameters and also the pre-stretch factor on the deformation behavior of the oriented tape (under each loading angle) was realized. Based on the physical meaning of the parameters and the observations in the parameter sensitivity analyses, the set of Hill parameters and prestretch factor, as given in Table 8, is adopted. As for the elastic and viscoplastic properties of the amorphous phase, the Hill parameters and prestretch factor are also identified based on the experimental results under strain rates of $\dot{\varepsilon} = 10^{-4}$ s$^{-1}$ and $\dot{\varepsilon} = 10^{-3}$ s$^{-1}$. The identified properties are then used to capture the stress-strain curves at strain rate of $\dot{\varepsilon} = 10^{-2}$ s$^{-1}$. Simulations are conducted for the oriented tape on an aggregate of 200 inclusions. Figure 10 shows the simulation results along with the experimental stress-strain curves.

Adequate agreements are obtained between the model predictions and experimental results. Although in some cases
$\varepsilon = 10^{-2} \text{ s}^{-1}$, the model predictions do not match the experimental results perfectly, comparison with Figure 7 still shows a significant improvement of the model predictions. Also, the dominant role of the anisotropy of the amorphous phase is noticeable. It was observed that the initially strongly oriented crystallographic texture further orients with the chain direction toward the drawing axis during deformation. Figure 11 shows the value of the von Mises effective stress of the amorphous phase and the crystalline lamella of each composite inclusion at the end of the deformation, under a strain rate of $\varepsilon = 10^{-2} \text{ s}^{-1}$ at different loading angles. The values of the effective stress are shown in pole figures which depict the distributions of the lamellar normals.

Clearly, by increasing the loading angle, the effective stress of both phases decreases. For loading angles of $\phi = 0^\circ$ and $\phi = 20^\circ$, stress of the amorphous phase and the crystalline lamella of each composite inclusion at the end of the deformation, under a strain rate of $\varepsilon = 10^{-2} \text{ s}^{-1}$ at different loading angles. The values of the effective stress are shown in pole figures which depict the distributions of the lamellar normals.

FIGURE 12 The normalized plastic slip rates of the chain and transverse slip systems of the oriented tape at the yield point and under a strain rate of $\varepsilon = 10^{-2} \text{ s}^{-1}$, at loading angle $\phi = 0^\circ$ depicted in pole figures which represent the distributions of the lamellar normals. [Color figure can be viewed at wileyonlinelibrary.com]

FIGURE 13 The normalized plastic slip rates of the chain and transverse slip systems of the oriented tape at the yield point and under a strain rate of $\varepsilon = 10^{-2} \text{ s}^{-1}$, at loading angle $\phi = 50^\circ$ shown in pole figures that depict the distributions of the lamellar normals. [Color figure can be viewed at wileyonlinelibrary.com]
the amorphous phase of the inclusions which are less aligned with the loading direction (higher angle between the lamellar normal and loading direction), have higher stresses. It is also seen that for ϕ = 20° and also ϕ = 0°, the inclusions with large amorphous stresses are differently oriented than inclusions with large crystalline stresses. Figures 12 and 13 show the plastic shear rate of the chain slip and transverse slip systems at the yield point, normalized by the applied strain rate (\(\dot{\varepsilon} = 10^{-2} \text{ s}^{-1}\)), in pole figures that show the distributions of the lamellar normal for loading angles of ϕ = 0° and ϕ = 50°, respectively.

The transverse slip systems for the sample taken along the draw direction of the oriented tape (ϕ = 0°), are less active compared to the chain slip systems. Also, it can be observed that for ϕ = 0° the slip system (100)[001] is the most dominant one.

In contrast to a loading of ϕ = 0°, the activity of the transverse slip systems is higher than the chain slip systems for a loading angle of ϕ = 50°. It can also be seen that for ϕ = 50° one chain slip system, (010)[001], and one transverse slip system, (100)[010], are not active.

CONCLUSIONS

The objective of this study was to model and capture the anisotropic mechanical behavior of oriented PE within a micromechanical approach. Drawn PE tapes show a highly anisotropic and highly rate-dependent behavior. There are different sources of anisotropy in these materials. As discussed by Sedighiamiri et al. and discussed in the section, considering the anisotropic behavior and oriented distribution of the crystalline lamellae does not suffice to accurately predict the mechanical behavior of oriented PE tapes. Accordingly, in this study, an anisotropic model for the amorphous phase was incorporated in the micromechanical model. Anisotropic elasticity, internal stress into the elastic network and an anisotropic visco-plastic flow were included for this purpose. The modified model adequately predicts the experimental results and remarkable improvements were obtained compared to the original (partially isotropic) model predictions. It can be concluded that the effect of the anisotropy of the amorphous phase in oriented PE tapes is significant and it can not be neglected.

ACKNOWLEDGMENTS

The authors gratefully acknowledge ExxonMobil corporation for financial support of this study.

REFERENCES
7 S. Nikolov, I. Doghri, Polymer 2000, 41(15), 1883.
17 A. Bilby, L. R. T. Lardner, A. N. Stroh, Actes du IXe congrés international de mécanique appliquée (Bruxelles), 1956, vol. 8, p. 35.
30 D. J. A. Senden, G. W. M. Peters, L. E. Govaert, J. A. W. van Dommelen, Polymer 2013, 54(21), 5899.
31 D. J. A. Senden, S. Krop, J. A. W. van Dommelen, L. E. Govaert, Polymer 2012, 50(24), 1680.
34 G. Sachs, Z. VDI 1928, 72, 734.