Characterizing and modeling the progressive damage of off-axis thermoplastic plies: effect of ply confinement

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Abstract

Although the mechanisms of degradation are, overall, quite similar between thermoset and thermoplastic composites, thermoplastic laminates exhibit a higher ductility. Because of that, classical models developed for thermosets fail in predicting accurately the response of thermoplastic-based laminates.

Here, we demonstrate that simulations based on the properties identified on isolated thermoplastics unidirectional plies, as usually done for thermosets, result in large errors. This is explained by the significant load-bearing capacity of the ply in the transverse direction that remains even after the first-ply failure, as a result of the confinement provided by adjacent plies. This effect also exists in thermosets, but it is usually limited due to the brittle nature of the resin. We are improving the mesoscale damage modeling in the off-axis direction by introducing this confinement effect in an original way. We are using a pragmatic approach consisting of separating the progressive damage into two parts called the “diffuse damage regime” and the “transverse-cracking regime”, which we describe by two distinct damage parameters. We show that the proposed method accurately predicts the nonlinear damage behavior of the off-axis laminates, and is a viable path to tailor the classical mesoscale damage model to glass fiber-reinforced thermoplastic composites.

Keywords: mesoscale, damage, thermoplastic composites, off-axis, transverse damage

1. Introduction

Laminated composites have been widely used to fabricate structures for aerospace, automotive, and portable electronics industries. During the design phase of these structures, modeling and predicting the degradation up to failure is necessary to optimize the geometry and weight of the final products. Such degradation can be described at different length-scales, depending on the modeling frameworks, \textit{i.e.} microscale (fiber-level modeling), mesoscale (ply-level modeling), and macroscale (structural-level modeling) \cite{1, 2, 3, 4, 5}. Bridging between scales usually relies on homogenization of the quantities (\textit{e.g.} stress or

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strain) at a lower scale using a specific homogenization technique in order to obtain constitutive parameters at a higher scale [6, 7].

Mesoscale modeling received considerable attention, due to its success in predicting various damage mechanisms at a coupon or plate level [8]. At the mesoscale, a composite ply is considered as a homogeneous material with orthotropic material properties. A multidirectional laminate is usually modeled by stacking damageable plies with different material orientations, connected by cohesive elements to account for delamination. The damage modeling at this level enables one to analyze the development of the main damage modes e.g. diffuse matrix damage, transverse crack, fiber breakage, and delamination, with high accuracy and without sacrificing the computational cost [9].

Various mesoscale damage modeling strategies for analyzing the progressive degradation of laminated composites have been proposed over the last four decades [8, 10]. These modeling strategies may incorporate first-ply failure criteria [11, 12] that are coupled with linear or exponential softening rules [13, 14, 15, 16, 17], or utilize strain energy-based continuum damage mechanics (CDM) [8, 10, 18]. The capability of various mesoscale damage modeling strategies has been critically examined within a series of World Wide Failure Exercise (WWFE I, II, and III) [19, 20, 21]. In these exercises, the originators of various damage/failure models were invited to blindly predict failure behavior (e.g. maximum stress, failure strain, stress-strain curve, and transverse crack density evolution) of composite laminates, for several test case scenarios. The results of these exercises show that the prediction of progressive nonlinear failure of laminated composites is still a challenge, especially for laminates with off-axis plies, for which the response is not only guided by the stiffness of the fibers but dominantly affected by the matrix and fiber dispersion. Progressive mechanisms such as fiber/matrix debonding, matrix micro-cracking, or transverse cracking may largely influence the overall stiffness of the laminate. This is even more noticeable for glass fiber-reinforced composites, either thermoplastic- or thermoset-based ones, where the ratio between the stiffness along the fiber and transverse directions is relatively lower ($E_{11}/E_{22} = 7.7$ for typical glass/polypropylene [22], and $E_{11}/E_{22} = 3.5$ for typical glass/epoxy [23]) than for carbon fiber-reinforced composites ($E_{11}/E_{22} = 14.9$ for typical carbon/epoxy [24]).

An accurate description of the accumulation of transverse cracks is essential, and it has been extensively studied at the microscale [25] and mesoscale [6]. Transverse cracking displays distinct well-known features such as thickness dependency [26, 27, 28]. It is also well-known that a difference exists in cracking kinetics between inner plies (fully embedded in a laminate) and skin plies (that typically behave as inner plies with double thickness) due to the different confinement effects of the surrounding plies [25, 29, 30]. The role of the so-called “confinement effect”, i.e. how the rest of the stacking sequence influences the damage response of the ply, has also been demonstrated for the analysis of transverse cracking accumulation [23, 31, 32]. Ladevèze and Lubineau [30] showed that a minimum influence of stacking sequence on the damage of a single-ply had been observed (i.e. in the order of a few percents, this being a major achievement as this conclusion justifies the concept of a mesoscale model). Such a conclusion is valid when comparing different confined off-axis plies with different stacking sequences. However, the behavior of transverse cracking in an isolated ply (unidirectional (UD) laminates that are usually used for identification) and in a
confined ply (the same ply constrained by other plies) is very different because the contrast in confinement between the two situations becomes too large. In an isolated UD ply, the first transverse crack, which is usually governed by the inter-fiber distance and defect [2], results in a complete transverse failure. Meanwhile, when the ply is located inside a multi-directional laminate, the ply could accommodate the multiplication of transverse cracks before it reaches its maximum load [33]. Such a different behavior plays a significant role in determining the non-linearity in the stress-strain curve of the transverse stiffness guided laminates. Thus, the damage behavior identified in UD laminates totally ignores the effect of the transverse crack multiplication of this ply when embedded in a laminate. Hence, our objective is to propose a pragmatic strategy to capture this effect, as it becomes a major effect in thermoplastic laminates as they can accommodate many cracks before final failure.

Here, we propose a pragmatic mesoscale damage modeling strategy to improve the classical mesoscale damage model as an attempt to capture the effect of ply confinement. The classical model was formulated based on the LMT-Cachan’s mesoscale damage model [8, 34]. To account for the confinement effect on the transverse cracking, we separated the modeling of matrix damage into two parts, namely “diffuse damage regime” and “transverse-cracking regime”. The two regimes are described by two distinct damage parameters. In this paper, all identifications and simulations were tailored towards the modeling of continuous glass fiber-reinforced polypropylene (GF/PP). We performed several mechanical tests on various lay-ups to calibrate the material parameters required for the model. We validated our model predictions against the experimental tensile test results on (±45/90\textdegree)\textsubscript{s} laminates. The validation using a (±45/90\textdegree)\textsubscript{s} lay-up was deemed appropriate as this lay-up exhibited a strong non-linear stress-strain response and, thus, truly examining the capability of the model contrary to more typical lay-ups (e.g., quasi-isotropic lay-up of (0/±45/90)\textsubscript{ns} containing a circular hole). The quasi-isotropic lay-up is indeed often used in the validation of recent works [34, 35, 36] despite it features an almost perfectly elastic response up-to-failure, that can be easily predicted and does not discriminate between the different modeling approaches.

2. Formulation of mesoscale damage model

2.1. Strain energy density and thermodynamic damage force

Adopted from the original work of Ladevèze and coworkers [8, 34], we used the three-dimensional (3D) strain energy of damageable ply, \( e_D \), written as follows.

\[
\epsilon_D = \frac{1}{2} \left[ \begin{array}{ccc} \sigma_{11} & \sigma_{12} & \sigma_{13} \\ \sigma_{21} & \sigma_{22} & \sigma_{23} \\ \sigma_{31} & \sigma_{32} & \sigma_{33} \end{array} \right]^T \Omega \left[ \begin{array}{ccc} \sigma_{11} & \sigma_{12} & \sigma_{13} \\ \sigma_{22} & \sigma_{22} & \sigma_{23} \\ \sigma_{33} & \sigma_{33} & \sigma_{33} \end{array} \right] + \frac{\sigma_{12}^2}{2G_{12}^0(1-d_{12})} + \frac{\sigma_{13}^2}{2G_{12}^0(1-d_{13})} + \frac{\sigma_{23}^2}{2G_{23}^0(1-d_{23})} \tag{1}
\]

where \( \sigma_{ij} \) and \( d_{ij} \) are the stress components and damage variables, respectively. \( i = 1, 2, 3 \) represents fiber, in-plane transverse, and out-of-plane transverse directions, respectively.
Note that we assumed \( d_{23} = d_{12} \). The compliance \( \Omega \) can be written as follows.

\[
\Omega = \begin{bmatrix}
\frac{1}{E_{11}(1-d_{11})} & -\frac{\nu_{12}^0}{E_{11}} & -\frac{\nu_{23}^0}{E_{22}} \\
-\frac{\nu_{12}^0}{E_{11}} & \frac{E_{22}^0(1-\sigma_{22}^2)\sigma_{22}^2}{E_{11}} & 1 \\
-\frac{\nu_{23}^0}{E_{22}} & 1 & \frac{E_{22}^0(1-\sigma_{33}^2)\sigma_{33}^2}{E_{22}}
\end{bmatrix}
\]

where \( [a]_0^+ = 1 \) if \( a \geq 0 \); else \( [a]_0^+ = 0 \).

(2)

The damage evolution is guided by the thermodynamic damage force \( (Y_m) \), that is a superposition of the shear damage force \( (Y_{12}) \) and transverse damage force \( (Y_{22}) \). \( Y_m, Y_{12} \) and \( Y_{22} \) can be written as follows.

\[
Y_m = \max_{\tau<\ell} (Y_{12}(\tau) + b_Y Y_{22}(\tau))
\]

(3)

\[
Y_{12} = \frac{\partial e_{D}}{\partial d_{12}} \bigg|_{\tilde{\sigma},d_{22}} = \frac{1}{2(1-d_{12})^2} \left( \frac{\sigma_{12}^2 + \sigma_{13}^2 + \sigma_{23}^2}{E_{11}^0 + E_{22}^0 + E_{23}^0} \right)
\]

(4)

\[
Y_{22} = \frac{\partial e_{D}}{\partial d_{22}} \bigg|_{\tilde{\sigma},d_{12}} = \frac{1}{2(1-d_{22})^2} \left( \frac{\langle \sigma_{22}^2 \rangle + \langle \sigma_{33}^2 \rangle}{E_{22}^0 + E_{23}^0} \right)
\]

(5)

where \( b_Y \) is a material parameter controlling the respective influence of the shear and transverse damage modes on the total damage force. Note that the operator \( \max_{\tau<\ell}(O(\tau)) \) has been used to make sure that the thermodynamic damage force \( Y_m \) can only monotonically increase (no damage healing allowed).

2.2. Classical mesoscale damage model

In the original publication [8], the linear damage evolution is used to appropriately describe the behavior of thermoset-based composites. Here, we used an exponential damage evolution because it fits better the experimental results on the investigated material i.e. glass fiber-reinforced polypropylene (GF/PP). Details on materials and processing can be found in subsection 3.1. The evolution of the exponential shear damage \( d_{12} \) can be written as follows.

\[
d_{12} = \min \left\{ d_{\text{slope}}, \left( 1 - \exp^{-\frac{1}{d_{\text{slope}}}} \right), 1 \right\} \text{ if } Y_m < Y_{m,\text{crit}}^{tc}; \text{ else } d_{12} = 1
\]

(6)

where \( d_{\text{slope}} \) and \( d_{\text{sate}} \) are model parameters for controlling the growth rate of the shear damage and its saturated value, respectively; \( Y_0 \) and \( Y_{m,\text{crit}}^{tc} \) are the thermodynamic thresholds corresponding to the shear damage initiation and complete failure, respectively. In such a model, the final failure of the ply under shear usually corresponds to the saturation of transverse cracking. This explains the notation \( Y_{m,\text{crit}}^{tc} \) for the threshold, in which “\( \text{crit} \)” stands for “critical” and “\( tc \)” stands for “transverse cracking”. These parameters were identified based on the stress-strain curves obtained from the monotonic tensile test on (±45)\(_{ns}\) laminates.

The evolution of the transverse tension damage \( d_{22} \) is dependent on the shear damage, \( d_{12} \), by a proportionality factor \( b_D \), shown below.

\[
d_{22} = \min \{ b_D d_{12}, 1 \} \text{ if } b_Y Y_{22} \leq Y_{m,\text{init}}^{tc}; \text{ else } d_{22} = 1
\]

(7)
where $Y_{tc_{m,init}}$ is a brittle damage threshold. In this model, the transverse damage $d_{22}$ would directly jump to a maximum value of 1 once $b Y_{22} \geq Y_{tc_{m,init}}$. $Y_{tc_{m,init}}$ corresponds to the threshold for the initiation of transverse cracks. It is accepted in the classical mesomodel that this initiation results in the immediate degradation of the transverse properties of the ply. Such a jump in damage variable may lead to a convergence difficulty and checkerboard damage patterns [37]. We refer to the above damage pattern as “classical model”, here onward. Note that such an approach is only suitable for modeling an isolated UD ply under transverse tension where only a single transverse crack triggers the final failure, or for simulating laminated sequences that do not include off-axis plies with multiple transverse cracks. In the case of a multi-directional laminate, the cracked ply could still sustain a significant level of load-bearing capacity even after the first transverse cracks occurs (see Fig. 1). In this setup, the multiplication of transverse cracks could be accommodated until the laminate finally reaches its maximum load-bearing capacity. Despite recent improvements proposed by enriching the damage mesomodel with micromechanics-inspired concepts and evolution laws [30, 29], these have not been transferred so far in an advanced and pragmatical mesoscale damage model for thermoplastic-based laminates. This motivates us to propose a modified version of the classical model, which we detail in the following subsection 2.3.

2.3. Proposed mesoscale damage model

To overcome the problem explained in the previous section, our pragmatic approach is to separate the progressive damage into two regimes, namely “diffuse damage-guided regime” and “transverse cracking-guided regime”. In this case, the proportionality coefficient between the shear damage force and the transverse tension damage force is differentiated in the “diffuse damage-guided regime” and in the “transverse cracking-guided regime” defined
as:

\[
\text{if } Y_m \leq Y_{m,\text{init}}^{\text{tc}}: \quad b_Y = b_Y^{\text{diff}} \quad \text{diffuse damage guided regime} \quad (8)
\]
\[
\text{if } Y_m > Y_{m,\text{init}}^{\text{tc}}: \quad b_Y = b_Y^{\text{tc}} \quad \text{transverse cracking guided regime} \quad (9)
\]

where \(Y_{m,\text{init}}^{\text{tc}}\) is the threshold separating the diffuse damage guided regime from the transverse cracking guided regime.

In the proposed approach, the shear damage evolution was modeled as in the classical approach. Indeed, the \((\pm 45)_{ns}\) tests used to identify the shear damage evolution law already accounts for both mechanisms, as 45° angle-ply traction involves not only a large development of matrix cracking and fiber/matrix debonding (diffuse damage) but also a large sequence of transverse cracking multiplication [38, 39, 29]. The difference lies in the way we tackle the modeling of transverse damage. The transverse damage \(d_{22}\) is built as a superposition between one contribution from the diffuse damage \(d_{22}^{\text{diff}}\) and one contribution from the transverse cracking \(d_{22}^{\text{tc}}\) as follows:

\[
d_{22} = d_{22}^{\text{diff}} + d_{22}^{\text{tc}}
\]

\[d_{22}^{\text{diff}} = \min \left\{ b_D^{\text{diff}} d_{12}, 1 \right\}; \quad d_{22}^{\text{tc}} = 0\] (11)

\[d_{22}^{\text{diff}} = \min \left\{ b_D^{\text{diff}} d_{12}^{\text{tc,init}}, 1 \right\}; \quad d_{22}^{\text{tc}} = \min \left\{ b_D^{\text{tc}} (d_{12} - d_{12}^{\text{tc,init}}), 1 \right\}\] (12)

\[d_{22}^{\text{diff}} = 1; \quad d_{22}^{\text{tc}} = 1\] (13)

where \(b_D^{\text{diff}}\) and \(b_D^{\text{tc}}\) are the proportionality factors between the transverse and shear damages that can vary with the regime.

\(d_{12}^{\text{tc,init}}\) is the level of the diffuse shear damage at the transition between both regimes defined as:

\[
d_{12}^{\text{tc,init}} = d_{12} | Y_m = Y_{m,\text{init}}^{\text{tc}}\] (14)
In this study, we suggested identifying $Y_{m,init}^{tc}$ by fitting the failure strain of $(90)_n$ under a tensile test.

**Remark:** Note that the above formulation considers only thick-ply cases (thickness $\geq 2h$, with $h$ the thickness of the prepreg) where the failure is usually guided by the stresses, and hence thickness-independent. In the case of thin-ply laminates, it is advised to modify $Y_{m,init}^{tc}$ as a function of thickness [6, 23], which is out of the scope of the work presented here.

Fig. 2 represents the shear and transverse damage evolution based on the classical and proposed damage models. One can see the proposed model as the regularized version of the classical model to account for the progressive strength distribution during transverse crack multiplication, which cannot be modeled using the classical approach.

### 2.4. Plasticity modeling

In addition to damage, the non-linearity of the global response of the off-axis composite laminate is, for a large part, caused by the plastic deformation of the polymeric matrix. Therefore, to allow a good validation with the experimental results, it is necessary to combine the damage model with a plastic yield function $f$, as shown below.

$$f = \sqrt{\tilde{\sigma}_{12}^2 + \tilde{\sigma}_{13}^2 + \tilde{\sigma}_{23}^2 + \alpha^2(\tilde{\sigma}_{22}^2 + \tilde{\sigma}_{33}^2) + \beta(\tilde{\sigma}_{22} + \tilde{\sigma}_{33}) - \sigma_y}$$  \hspace{1cm} (15)$$

where the yield strength $\sigma_y$ of the GF/PP is fitted by a power law function as defined below.

$$\sigma_y = A + B (\bar{\varepsilon}^p)^C$$  \hspace{1cm} (16)$$

The yield function $f$ is expressed in terms of the effective stress $\tilde{\sigma}_{ij} = \sigma_{ij}/(1 - d_i)$ to account for the effect of damage on the plastic yielding. Here, $d_i$ is the corresponding damage variable for each directions, $\bar{\varepsilon}^p$ is the scalar equivalent plastic strain to be solved by satisfying consistency conditions (i.e. $\bar{\varepsilon}^p$ is the solution of $f = 0$, and $\dot{f} = 0$ during plastic yielding). Note that $\sigma_{11}$ term was removed from the yield function $f$ because we assumed that there was no plastic strain along the fiber direction. Taking into account $\sigma_{11}$ term into the plastic yield function may induce strong anisotropy and leads to a convergence issue since the order of $\sigma_{11}$ is much higher than other stress terms. In addition, we adopted a pressure-dependent yield function to account for different yield strengths of laminate under tension and compression, which was not accounted in the original publication of the LMT-Cachan model [8]. However, the inclusion of a pressure term into the yield function of anisotropic materials often introduces an unrealistic plastic strain [40]. This can be shown, for example, in the case of pure shear loading ($\tilde{\sigma}_{12} \neq 0, \tilde{\sigma}_{11} = \tilde{\sigma}_{22} = \tilde{\sigma}_{33} = \tilde{\sigma}_{13} = \tilde{\sigma}_{23} = 0$), parasitic plastic strain in the transverse direction will be introduced ($\varepsilon^p_{22} \neq 0$) because $\frac{\partial f}{\partial \tilde{\sigma}_{22}} \neq 0$. In order to rectify this issue, we calculated the plastic flow direction based on a non-associative flow function $g$, which does not include the pressure term as shown below.

$$\dot{\varepsilon}^p = \dot{\varepsilon}^p \frac{\partial g}{\partial \tilde{\sigma}} \text{ where } g = \frac{\sqrt{\tilde{\sigma}_{12}^2 + \tilde{\sigma}_{13}^2 + \tilde{\sigma}_{23}^2 + \alpha^2(\tilde{\sigma}_{22}^2 + \tilde{\sigma}_{33}^2)}}{7}$$  \hspace{1cm} (17)$$
where \( \dot{\varepsilon}_p \) and \( \frac{\partial g}{\partial \sigma} \) are the magnitude and direction of the plastic strain increment. Using such non-associative flow function enables one to include a pressure-dependent yield strength without introducing an unrealistic plastic strain.

2.5. Viscous regularization for localization limiter and convergence-aid

In the classical continuum damage mechanics, energy dissipation due to damage is sensitive to the element size due to the localization phenomenon. This creates a spurious mesh-dependent problem. In this work, we used a viscous regularization technique to regularize the damage variables to obtain mesh-objective simulations. Such regularization also improves the convergence speed by providing a positive tangent stiffness for a sufficiently small time-step. In this work, we adopted a simple viscous regularization, as shown below.

\[
d_{i,v}^{[n+1]} = \left( \frac{\Delta t}{\Delta t + \tau_r} \right) d_{i,v}^{[n+1]} + \left( \frac{\tau_r}{\Delta t + \tau_r} \right) d_{i,v}^{[n]} \quad (18)
\]

where the subscript \( i \) can be replaced by 11, 22, 12 representing variables associated to fiber-, transverse- and shear damages, respectively. \( d_i \) is the actual damage value as a direct function of local strain at time \( t^{[n+1]} \), \( d_{i,v} \) is the regularized damage value and \( \tau_r \) is regularization constant. Unlike the delay-effect technique used in the original publication [41], the regularized damage \( d_{i,v}^{[n+1]} \) can be obtained directly without iterations and therefore, it speeds up computational time.

3. Experimental setup and results

3.1. Materials and processing

We applied the proposed model to predict the non-linear plastic and damage behaviors of thermoplastic-based composites. However, it should be noted that the model is also applicable to thermoset-based composites (e.g., carbon/epoxy or glass/epoxy). Here, the material under investigation is a continuous glass fiber-reinforced polypropylene (GF/PP) composite produced by SABIC. The laminated composite is composed of E-glass fibers with an average diameter of 16.3 \( \mu \text{m} \) and a volume fraction of 41.6 \( \pm \) 2.6\%; the matrix is an impact-grade copolymer PP. The elementary GF/PP ply was provided to us in the form of a UD tape with a ply thickness of 255 \( \mu \text{m} \). We stacked the UD-tapes in accordance with the specified orientations and put them into a steel mold. The mold was then compressed under a 7.5 bar pressure, using a static press machine from Pinette Emidécau Industries. Details on material processing are provided in Refs. [42, 22, 43].

3.2. Identification of material parameters

We performed a set of mechanical tests, listed in Table 1, required for the identification of the model parameters as well as for the validation of the model. Some parameters were assumed for the sake of simplicity. Fundamental background on the identification procedure for the mesoscale damage model, specifically the classical one, can be found on the original work of Ladevèze and LeDantec [8].
Table 1: Mechanical tests and composite lay-ups for calibration and validation of the constitutive model.

<table>
<thead>
<tr>
<th>Purpose</th>
<th>Lay-up</th>
<th>Thickness [mm]</th>
<th>Test type</th>
</tr>
</thead>
<tbody>
<tr>
<td>Calibration</td>
<td>(0)\textsubscript{4}</td>
<td>1</td>
<td>Monotonic tensile test at 10\textsuperscript{-3}/s</td>
</tr>
<tr>
<td>Calibration</td>
<td>(90)\textsubscript{8}</td>
<td>2</td>
<td>Monotonic tensile test at 10\textsuperscript{-3}/s</td>
</tr>
<tr>
<td>Calibration</td>
<td>(±60)\textsubscript{2s}</td>
<td>8</td>
<td>Monotonic tensile test at 10\textsuperscript{-3}/s</td>
</tr>
<tr>
<td>Calibration</td>
<td>(±45)\textsubscript{s}</td>
<td>1</td>
<td>Monotonic tensile test at 10\textsuperscript{-3}/s</td>
</tr>
<tr>
<td>Calibration</td>
<td>(±45)\textsubscript{s}</td>
<td>1</td>
<td>Load/unload tensile test at 10\textsuperscript{-3}/s</td>
</tr>
<tr>
<td>Validation</td>
<td>(±45/90)\textsubscript{2s}</td>
<td>4</td>
<td>Monotonic tensile test at 6 mm/min</td>
</tr>
</tbody>
</table>

We performed the identification of material parameters by comparing the finite element (FE) simulation results with those obtained from experiments. The dimension of FE models followed that of the test specimens in experiments. Note that in all identification and validation processes, $\Delta T$ of -100\degree C was applied to the model before imposing any mechanical loading; this is necessary to capture the residual stresses in the laminate after manufacturing correctly.

The identification was sequentially performed for each ply orientation. First, we started by getting the basic elastic material parameters ($E_{11}^0$, $E_{22}^0$, $G_{12}^0$, $\nu_{12}^0$) from tensile test of (0)\textsubscript{4}, (90)\textsubscript{8} and (±45)\textsubscript{s} laminates. Note that the single ply was considered as transversely isotropic for the sake of simplicity.

We then identified the shear damage parameters ($d_{sat}$, $d_{slope}$, $Y_{0}$, $Y_{tc_{m,init}}$, $b_{Y_{tc}}^{diff}$, $b_{Y_{tc}}^{c}$) by decoupling the damage evolution (shear stiffness degradation) and plastic shear strain based on the results of the loading-unloading tensile test of (±45)\textsubscript{s} laminates. Fig. 3a and Fig. 3b show the shear damage evolution and plastic shear strain, respectively. $b_{Y_{tc}}^{diff}$ and $b_{Y_{tc}}^{c}$ are the coupling parameters between the shear and transverse tension thermodynamic damage forces. Here, we assumed that $b_{Y_{tc}}^{diff} = b_{Y_{tc}}^{c} = 4$ based on the results of a previous study that shows $G_{2c} = 4G_{1c}$ [44]. The identification of $b_{Y_{tc}}^{diff}$ from the (±60)\textsubscript{s} laminate was found ineffective, due to its insensitivity towards $b_{Y_{tc}}^{diff}$.

Our next step was to identify the plasticity parameters ($A$, $B$, and $C$) in Eq. 16 by performing a power-law fitting on the plastic shear strain $\sigma_y$ vs. yield stress $\bar{\epsilon}_{vp}$ data obtained from the (±45)\textsubscript{s} loading-unloading test. To do so, it was necessary to decouple the damage and plasticity contribution, as previously explained.

We then calibrated the transverse damage parameters ($b_{D_{tc_{m,init}}}^{diff}$, $Y_{m,init}^{tc}$, $\alpha$) by fitting stress-strain curves of (90)\textsubscript{8} plies under transverse tension. $b_{D_{tc}}^{diff}$ and $\alpha$ controlled the non-linearity of the (90)\textsubscript{8} behavior due to the diffuse damage and plasticity, respectively, and $Y_{m,init}^{tc}$ governed the failure point on (90)\textsubscript{8} plies. In this case, we assumed the pressure-dependent parameter $\beta$ to be 5 MPa for simplicity [45]. Since most test cases of interest correspond to tensile loading, the value of $\beta$ is insignificant and does not affect the results of the analysis.

The last step is to calibrate one of the most critical parameters in this study, i.e. $b_{D}^{c}$, which defines the coupling between the transverse tension damage and the shear damage during transverse cracks multiplication regimes. $b_{D}^{c}$ was calibrated by fitting the model...
prediction against the tensile stress-strain curve of \((\pm 60)\) laminates. We found that 3.5 was the value suitable for our thermoplastic-based composites. It is worth mentioning that another way to obtain \(b_{tcD}\) is to use the micro-to-mesoscale relationship defined by Lavedeze and Lubineau [30, 46, 29]. In their work, the model was tailored towards thermoset-based composites (carbon/epoxy), and the approximated value for \(b_{tcD}\) is within a 1.5-2.5 range. However, it should be noted that their approach could only give the order of magnitude and highly dependent on the uncertainties related to the morphology of the cracking pattern, microstructures, and plastic behavior of the matrix.

3.3. Mechanical tests for validating the proposed model

We validated the prediction of the proposed model against discriminant mechanical test results. We conducted a tensile test on GF/PP laminates with a lay-up of \((\pm 45/90)_{2s}\) at a loading speed of 6 mm/min. This speed was chosen to ensure that the local strain rates in the specimen were within the calibration range \((0.001/s to 0.1/s)\). The selected lay-up was deemed suitable because it exhibited a strong non-linearity in the stress-strain curve, due to matrix-dominated damages. The specimen was 110 mm-long, 20 mm-wide, 4 mm-thick (ASTM Standard D7264). The tensile test was performed using an Instron 5882 with a 100 kN load-cell; the strain field on the surface of the specimen was measured using a digital image correlation (DIC) technique.

4. Results and discussions

4.1. Calibration of mesoscale parameters

Fig. 4a shows the calibration results of the model in describing the behavior of the composite ply along 0° direction. In our study, the model was calibrated to capture the linearity of stress-strain curves, and failure point of the 0° UD plies correctly. Note that
the proposed model focused on the prediction of off-axis ply behavior. Nevertheless, we included $0^\circ$ calibration result for the sake of completeness. Fig. 4b shows the identification result of the model in describing the behavior of $(\pm 45)s$ laminates. The model capability in capturing the non-linearity of the stress-strain curve (arising from damage and plasticity) as well as the failure onset has been successfully calibrated.

The calibration result for the behavior of $90^\circ$ UD plies is shown in Fig. 5a. The stress-strain behavior is initially linear, followed by a slight nonlinearity due to plasticity, and ended by brittle failure. The matrix cracking that dominates such brittle behavior leads to a transverse crack and is governed by the nearest inter-fiber distance and defect [2]. In a UD case, $90^\circ$ UD plies can only allow for a single transverse crack. Therefore, modeling of such brittle behavior is suitable for an isolated case of $90^\circ$, but not when the ply is embedded within a laminate or attached as a skin ply. As explained previously, when $90^\circ$ plies are placed inside a laminate or as a skin ply, it can still sustain a remarkable amount of stress when continued to be loaded even after the presence of the first transverse crack. The confinement effect of the adjacent plies results in higher stress-bearing capability. Fig. 5b shows the behavior predicted by the proposed model of $(90)_8$ UD plies using different values of $b_D^{tc}$. Modeling such behavior is vital since the transverse cracks, in reality, occur progressively. This important effect is clearly shown in the validation section by comparing the classical and proposed models. In this stage, $b_D^{tc}$ has been identified by fitting the tensile test results of $(\pm 60)_2s$ laminates, as shown in Fig. 6a. The behavior of $(\pm 60)_2s$ laminates is linear-elastic, followed by a relatively flat plateau contributed by progressive damage due to shear and transverse damage. Here, we choose the value of $b_D^{tc} = 3.5$ to fit the experimental curve. Note that we study only the strain range from 0% up to 4% since our research scope is limited to a small strain framework. This framework is also reasonable because the localization and final failure of the sample occur around 4.5 - 5% strain, as shown in the DIC image in Fig. 6b.
Figure 5: Identification of parameters based on monotonic tensile tests of (90°)$_8$ at 0.001/s: (a) stress-strain curve up to final failure, (b) stress-strain curve of UD (90°) plies when they are confined by adjacent plies; effect of varying \( b_{tc} \) on the softening and stiffening behaviors after the first transverse cracking.

Figure 6: (a) Identification of \( b_{d,tc} \) by fitting the tensile stress-strain response of (±60)$_2$$_s$ laminates, (b) strain contour at different levels obtained by DIC technique.
Table 2: Calibrated material parameters for mesoscale modeling of glass/polypropylene laminates.

<table>
<thead>
<tr>
<th>Parameter</th>
<th>Value</th>
<th>Unit</th>
<th>Remarks</th>
</tr>
</thead>
<tbody>
<tr>
<td><strong>Elasticity</strong></td>
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<td></td>
<td></td>
</tr>
<tr>
<td>$E_{11}^0$</td>
<td>34300</td>
<td>MPa</td>
<td>Longitudinal tensile modulus</td>
</tr>
<tr>
<td>$E_{22}^0$</td>
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<td>MPa</td>
<td>Transverse tensile modulus</td>
</tr>
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<td>$G_{12}^0$</td>
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<td>MPa</td>
<td>In-plane shear modulus</td>
</tr>
<tr>
<td>$\nu_{12}^0$</td>
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<td>-</td>
<td>In-plane Poisson’s ratio</td>
</tr>
<tr>
<td>$\nu_{23}^0$</td>
<td>0.40</td>
<td>-</td>
<td>Out-of-plane Poisson’s ratio</td>
</tr>
<tr>
<td><strong>Matrix-dominated damage</strong></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>$b_{Y}^{diff}$</td>
<td>4</td>
<td>-</td>
<td>Shear and transverse damage force coupling in diffuse regime</td>
</tr>
<tr>
<td>$b_{Y}^{tc}$</td>
<td>4</td>
<td>-</td>
<td>Shear and transverse damage force coupling transverse crack regime</td>
</tr>
<tr>
<td>$Y_{m,init}^{te}$</td>
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<td>MPa</td>
<td>Thermodynamic force for transverse crack initiation</td>
</tr>
<tr>
<td>$Y_{m,crit}^{te}$</td>
<td>3</td>
<td>MPa</td>
<td>Thermodynamic force at final failure of $(\pm 45)_s$</td>
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<td>$b_{P}^{diff}$</td>
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<td>-</td>
<td>Shear and transverse diffuse damage ratio</td>
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<td>$b_{P}^{tc}$</td>
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<td>Shear and transverse cracking damage ratio</td>
</tr>
<tr>
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<td>Thermodynamic force for diffuse damage initiation</td>
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</tr>
<tr>
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<td>Fitting parameters for exponential damage evolution</td>
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<tr>
<td><strong>Pressure-dependent plasticity</strong></td>
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<td></td>
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<td>MPa</td>
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<td>Anisotropy coefficient in yield strength</td>
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<td>Plastic strain hardening coefficient</td>
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<tr>
<td>$C$</td>
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<td>-</td>
<td>Plastic strain hardening exponent</td>
</tr>
</tbody>
</table>
4.2. Validation of the proposed model

In this subsection, we validated the predictions (i.e. stress-strain curves and damage patterns) given by the classical and proposed mesoscale damage models against the experimental results.

Fig. 7a shows the stress-strain curves predicted by both classical and proposed mesoscale damage models. The proposed model captures the nonlinearity of the experimental curves correctly, whereas the classical damage model is unable to do so. The prediction using the classical model produces a plateau in the stress-strain curve due to the presence of alternating damage patterns; such damage pattern was unable to provide stress hardening as in the real samples. This is contrary to the strain contour observed experimentally based on the DIC technique. Fig. 7b shows a uniform strain contour even in the transverse-crack guided regimes (no alternating damage pattern observed). The prediction using the classical model could be improved if a very fine mesh and the variability of material properties were accounted for in order to capture the progressive transverse cracking. Indeed, such an improved strategy would lead to a correct simulation of progressive transverse cracking, but the computational cost will increase significantly due to mesh refinement.

In our proposed approach, the effect of the progressive transverse cracks is captured by homogenizing their effect throughout the whole length of the laminate, using progressive softening behavior of 90° plies with a tailorable value of $b_{d,tc}$. Hence, our pragmatic approach can capture progressive behavior with a relatively coarse element size and, thus, computationally inexpensive in order to achieve an accurate prediction.

The influence of the element size on the modeling results is shown in Fig. 8a. Three element sizes of 1, 1.5, and 2 mm are introduced separately into the models. The elements are rectangular, and each ply is modeled by one element through-thickness. Fig. 8a reveals that the stress-strain curves generated by both classical and proposed models are generally mesh-independent. The final failure (displayed by a sudden load-drop) shows a slight
Figure 8: Effect of element sizes for both the classical model and the proposed model on (a) stress-strain curves and (b) deformed shapes with the shear damage contour.
mesh-dependency. Fig. 8b shows the damage pattern in the failed sample of \((\pm45/90_2)_2s\) laminates obtained from both classical and proposed models. The results of the classical model show an alternating damage pattern once \(Y_m > Y_{m,init}^{tc}\). This is triggered by a sharp jump of the damage value from \(d_{22}\) to 1. The use of excessive viscous regularization could overcome this alternating damage pattern, but the accuracy of the prediction would be compromised. Meanwhile, the proposed model could capture the failed sample from experiments quite well. The damage pattern obtained by the proposed model is free from alternating damage patterns since the damage evolution of 90° plies has been handled progressively. A localized failure zone is observed in both the simulation and the experiment. The difference in orientation (straight in simulation versus oblique in experiments) comes from the fact that we did not introduce interlaminar elements allowing delamination in the simulation.

5. Conclusions

Here, we propose a pragmatic progressive mesoscale damage modeling approach considering the effect of confining plies on the off-axis behavior of composite laminates. We modified the widely-used LMT-Cachan mesomodel by separating the matrix damage model into two regimes, the diffuse damage-guided regime, and the transverse crack-guided regime, and we assigned different proportionality constants to these two regimes. We calibrated the mesoscale material parameters required in the model using experiments. We applied the new approach to predict the non-linear response and damage pattern of glass-reinforced polypropylene laminates with the lay-up of \((\pm45/90_2)_2s\). We showed that our approach successfully captured the nonlinear stress-strain curves and damage pattern of the \((\pm45/90_2)_2s\) laminates. These results suggest that the proposed pragmatic approach is a viable path to tailor the classical mesoscale damage model for glass fiber-reinforced thermoplastic composites.

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Data Availability

Data will be made available on request.

References


