Progressive Failure Analysis of 3D Woven Composites via Multiscale Recursive Micromechanics

Trenton M. Ricks\textsuperscript{1}, Evan J. Pineda\textsuperscript{2}, Brett A. Bednarcyk\textsuperscript{3}, and Steven M. Arnold\textsuperscript{4}

\textit{NASA Glenn Research Center, Cleveland, OH, 44135, U.S.A}

Multiscale progressive failure simulations have been performed for a 3D woven composite considering six length scales, spanning from the woven composite repeating unit cell to the matrix material containing voids. The Multiscale Recursive Micromechanics approach, which enables micromechanics models to call other micromechanics models (or themselves recursively) to consider finer and finer length scales, has been employed. The multiscale model uses both the generalized method of cells and Mori-Tanaka micromechanics theories and considers failure and damage at the lowest length scale. A simple subvolume elimination damage model, as well as the crack band model, are employed for the matrix and their predicted composite responses compared with and without a preexisting binder tow disbond. Results are also compared with experimental data for an AS4 carbon fiber/RTM6 epoxy matrix 3D orthogonal woven PMC, with good agreement in terms of global stiffness and global failure stress.

I. Introduction

Three-dimensional (3D) woven carbon fiber reinforced polymer (CFRP) composites are emerging as a viable aerospace material because of their observed damage tolerance and enhanced through-thickness properties, as compared to traditional layered composites \cite{1}-\cite{11}. Further, structures containing thick sections and/or complex geometries can be fabricated using a single 3D woven preform that is infused with resin. However, many morphological features, such as voids, cracks, disbonds, wrinkling/waviness of tows, variability in cross section of tows and tow misalignment, can be observed within the microstructure of these composites after manufacturing \cite{9},\cite{12}. It is unclear whether these features should be considered defects that negatively affect the composite performance or if the overall response at the component scale is insensitive to the presence of these features. NASA’s Composites Technology for Exploration (CTE) project, and other projects, have focused on both experimental and analytical campaigns to better understand these materials and improve their technology readiness level (TRL) for replacing heavy metallic ring-frames used to join adjacent stages in launch vehicles \cite{12},\cite{13}.

Progressive failure analysis (PFA) of 3D woven composites remains a challenge due to the complex geometry and the extremely fine meshes required to capture accurately the local fields using finite element approaches. Thus, the extent of PFA simulations in the open literature is limited compared to more traditional composites \cite{14}. The trend in model development has been to increase the geometric fidelity of the model, which is on the critical path for development of a virtual testing platform for 3D woven composite structures \cite{15}-\cite{24}. However, the computational cost of these models can be exorbitant as many the models contain upwards of hundreds of thousands, or even millions, of degrees of freedom. Of equal importance is the development of rapid engineering tools for material and structural

\textsuperscript{1} Research Aerospace Engineer, Materials and Structures Division, AIAA Member.
\textsuperscript{2} Research Aerospace Engineer, Materials and Structures Division, AIAA Senior Member.
\textsuperscript{3} Research Materials Engineer, Materials and Structures Division, AIAA Associate Fellow.
\textsuperscript{4} Technical Lead: Multiscale Modeling, Materials and Structures Division, AIAA Senior Member.
design. With such tools, parametric and statistical studies can be conducted incorporating a variety of loading scenarios and levels of imperfections. Recently, the engineering constants for a 3D woven repeating unit cell (RUC) were predicted utilizing the multiscale generalized method of cells (MsGMC) and were compared to voxel FEM model and experimental data, showing reasonable agreement with both [12].

The focus on the current work is to develop and validate a multiscale modeling tool capable of ultra-efficient analysis of a 3D woven composite RUC containing defects, including voids and cracks. The NASA Multiscale Analysis Tool (NASMAT), developed at the NASA Glenn Research Center, is the multiscale platform used to conduct the analysis [25]. Within NASMAT, the multiscale recursive micromechanics (MsRM) framework, which is a generalization of the multiscale Generalized Method of Cells (MSGMC) [26], is employed. The MsRM technique enables micromechanics methods to call themselves in a recursive manner to capture the effects of the material microstructure at lower and lower length scales. The MsRM implementation benefits from the use of recursive data structures, enabling NASMAT to consider any arbitrary number of length scales. Furthermore, the implementation is micromechanics theory agnostic, meaning any micromechanics theory can operate at any length scale, being called by, or calling, other theories. Herein, the MsRM capabilities of NASMAT are applied to a 3D woven composite, which has multiple length scales (i.e., weave pattern made up of tows (yarns) and the microstructure within a tow consisting of fibers and matrix). As will be discussed, the MsRM approach tracks the stress and strain fields throughout the composite (according to the chosen micromechanics theory) at every length scale. As such, damage and failure can be predicted based on the local fields, and the effects can be homogenized to impact the material response at the higher length scales. To maintain ultra-efficiency in the computational cost, the generalized method of cells (GMC) and Mori-Tanaka (MT) micromechanics theories are used to model the constituents and RUCs at the various scales in the model [27],[28]. Furthermore, a modified crack band damage model based upon the work of refs [29]-[31], along with a simple subvolume elimination damage model, has been employed at the lowest length scale for the 3D woven composite for the matrix within and between the tows. Results compare these two damage modeling approaches for predicting the uniaxial stress-strain curves of the composite for each applied strain component (normal and shear). In addition, the impact of the crack band damage initiation stress of the matrix and the presence of a preexisting binder tow disbond are examined. Porosity of the matrix material has also been incorporated at the lowest length scale. By using efficient semi-analytical micromechanics theories (as opposed to finite element analysis) on each scale, MsRM can simulate the nonlinear progressive failure response of 3D woven composites extremely efficiently, even when considering multiple length scales.

II. Overview of Multiscale Recursive Micromechanics (MsRM)

The MsRM approach for modeling materials with multiscale microstructures is shown schematically in Figure 1 [32]. The approach relies on recursive procedures, subroutines, and data structures that allow an arbitrary number of scales to be defined and seamlessly pass data between each other. The approach admits any micromechanics theory that provides a strain concentration tensor \( A_{ij}^{(\alpha_i)} \), which relates the local strains within subvolumes in the material to the global strains, with the subvolumes (at a given scale, \( i \)) denoted in general by \( \alpha_i \), and the number of subvolumes by \( N_{\alpha_i} \). This concentration tensor provides the local strains in the subvolumes in terms of the average (global) strains, \( \bar{\varepsilon} \),

\[
\varepsilon_{ij}^{(\alpha_i)} = A_{ij}^{(\alpha_i)} \bar{\varepsilon}_i \tag{1}
\]

Substituting Eq. (1) into the local constitutive equation,

\[
\sigma_{ij}^{(\alpha_i)} = C_{ij}^{(\alpha_i)} \varepsilon_{ij}^{(\alpha_i)} \tag{2}
\]

gives,

\[
\sigma_{ij}^{(\alpha_i)} = C_{ij}^{(\alpha_i)} A_{ij}^{(\alpha_i)} \bar{\varepsilon}_i \tag{3}
\]

The average (global) stress tensor is given by,
Figure 1. Schematic of the Multiscale Recursive Micromechanics (MsRM) approach wherein micromechanics models are embedded within each other to model microstructure at any number of length scales [32].

\[
\bar{\sigma}_i = \sum_{\alpha=1}^{N_{\alpha}} v_\alpha \sigma_i^{(\alpha)} \tag{4}
\]

where \( v_\alpha \) is the volume fraction of subvolume \( \alpha \). Eqs. (3) and (4) lead to,

\[
\bar{\sigma}_i = \sum_{\alpha=1}^{N_{\alpha}} v_\alpha C_i^{(\alpha)} A_i^{(\alpha)} \bar{\varepsilon}_i \tag{5}
\]

and recognizing that the effective elastic constitutive equation at level \( i \) is given by,

\[
\bar{\sigma}_i = C_i^* \bar{\varepsilon}_i \tag{6}
\]

Eqs. (5) and (6) indicate that the effective stiffness tensor, \( C_i^* \), at level \( i \) is given by,

\[
C_i^* = \sum_{\alpha=1}^{N_{\alpha}} v_\alpha C_i^{(\alpha)} A_i^{(\alpha)} \tag{7}
\]

In MsRM, the scales are linked by equilibrating the homogenized average stress, strain, and stiffness tensors at Level \( i \) to the local stress, strain, and stiffness tensors of a given subvolume at Level \( i-1 \) (with appropriate transformation to account for the potential coordinate system change from scale to scale). That is,

\[
\bar{\varepsilon}_i = T_2^{(i)} \varepsilon_{i-1}^{(\alpha)}, \quad \bar{\sigma}_i = T_2^{(i)} \sigma_{i-1}^{(\alpha)}, \quad C_i^* = T_4^{(i)} C_{i-1}^{(\alpha)} \quad i = 1, \ldots, k \tag{8}
\]

where \( T_2^{(i)} \) and \( T_4^{(i)} \) are the appropriate second and fourth order coordinate transformation tensors, respectively. Hence, it is clear that starting with the lowest scale (\( k \)) microstructure (see Fig. 1), whose subvolumes contain only monolithic materials, the effective stiffness tensor can be calculated using any standard micromechanics theory. This stiffness tensor (after appropriate coordinate transformation) then represents the homogenized material occupying one of the subvolumes within a composite material at the next higher length scale. Given the transformed effective stiffness...
tensors of all subvolumes at this next higher length scale, the effective stiffness tensor of the composite at this level can be determined. This stiffness tensor can then be transformed and passed along to the next higher length scale, and the process repeats until the highest length scale considered (0) is reached.

As an example, for an MsRM analysis considering three length scales (0, 1, and 2), the overall effective stiffness tensor can be written using Eqs. (7) and (8) as,

\[
\mathbf{C}_0^{*} = \sum_{a_0} \mathbf{v}_{a_0} \left( \mathbf{T}_1^{a_1} \right)^{-1} \sum_{a_1} \mathbf{v}_{a_1} \left( \mathbf{T}_2^{a_2} \right)^{-1} \sum_{a_2} \mathbf{v}_{a_2} \mathbf{C}_2^{(a_2)} \mathbf{A}_2^{(a_2)} \mathbf{A}_1^{(a_1)} \mathbf{A}_0^{(a_0)}
\]

(9)

Note that in Eq. (9), the superscript on the bracketed terms indicates that all variables within the brackets are a function of the subvolume indices from the next higher length scale (including lower scale volume fractions and subvolume indices). The intent of this notation is to fully define the subvolume at a given scale as one progresses down the length scales. For example, using this notation, the effective stiffness tensor at Level 2, from Eq. (8), can be written as,

\[
\left\{ \left[ \mathbf{C}_2^{*} \right]^{(a_2)} \right\}^{(a_1)} = \left\{ \left[ \mathbf{T}_2^{a_2} \right]^{(a_1)} \right\}^{(a_0)} \left\{ \mathbf{A}_2^{(a_2)} \right\}^{(a_0)} \left\{ \mathbf{A}_1^{(a_1)} \right\}^{(a_0)} \left\{ \mathbf{T}_1^{a_1} \right\}^{(a_0)} \mathbf{A}_0^{(a_0)}
\]

(10)

as there are distinct \( \mathbf{C}_2^{*} \) values for every Level 1 subvolume, while there are distinct Level 1 composites present within each Level 0 subvolume.

Converse to this multiscale homogenization procedure, MsRM can perform multiscale localization of the stress and strain tensors. The multiscale localization is needed for inclusion of nonlinearity from damage (and inelasticity). For the three length scale example, the local strain tensor in an arbitrary lowest scale (level 2) subvolume can be written using Eqs. (1) and (8) as,

\[
\left\{ \left[ \mathbf{e}_2^{(a_2)} \right]^{(a_1)} \right\}^{(a_0)} = \left\{ \left[ \mathbf{A}_2^{(a_2)} \right]^{(a_1)} \right\}^{(a_0)} \left\{ \left[ \mathbf{T}_2^{a_2} \right]^{(a_1)} \right\}^{(a_0)} \left\{ \mathbf{A}_1^{(a_1)} \right\}^{(a_0)} \left\{ \mathbf{T}_1^{a_1} \right\}^{(a_0)} \mathbf{A}_0^{(a_0)} \mathbf{\hat{e}_0}
\]

(11)

Again, the superscript on the bracketed terms indicates that all variables within the brackets are a function of the subvolume indices from the next higher length scale. The stress tensor for any subvolume at any length scale can be similarly determined through localization, or by simply using the strain tensor, along with the constitutive equation (2), at the appropriate length scale. Note that, although not shown here for simplicity, the MsRM implementation in NASMAT includes thermal stresses as well. Because of its ability to handle multiple length scales in a single analysis, MsRM is ideal for multiscale modeling of materials such as 3D woven composites that exhibit identifiable microstructures across multiple length scales.

III. 3D Woven Composite Material

NASA’s recent CTE project is focused on developing 3D woven CFRP materials as a potential lighter-weight, damage tolerant, replacement for metallic ring-frames used to join the vertical stages of a launch vehicle. As shown in Figure 2, tests have been conducted at multiple length scales. Figure 2a shows an X-Ray CT image of a flat panel from which coupon specimens were cut and tested under tension and compression, room temperature/dry and elevated temperature/wet conditions, and single shear bearing [12]. In addition, acid digestion tests were conducted to ascertain the overall fiber volume fraction and void content. Figure 2b displays the C-channel tension/compression test specimen that was designed to represent a relevant structural configuration containing a key engineering feature in a joint design, i.e., a bend with a sharp radius. Finally, the CAD model of the largest test specimen considered, the C-Joint sub-element, is shown in Figure 2c. The C-joint sub-element contains a honeycomb sandwich panel (blue) which represents the acreage of the launch vehicle structure. This panel is bonded to the 3D woven C-rep flat section (magenta), which serves as a surrogate for the web in the C-channel, with a resin infused, 3D woven pi-preform and film adhesive (green). The C-joint sub-elements were tested at the NASA Marshall Space Flight Center (MSFC) under tension and compression. The specimen was mounted to an aluminum test stand and the load was introduced through an aluminum insert colored gray and yellow in Figure 2c, respectively.
Figure 2. Testing performed on 3D woven composite structures under NASA’s CTE project. (a) Flat panel coupons. (b) C-channel. (c) C-joint sub-element containing honeycomb sandwich panel (blue), 3D woven C-rep panel (magenta), and 3D woven pi-preform (green).

The CTE experimental program includes a thorough non-destructive evaluation (NDE) component which utilizes digital image correlation (DIC), X-Ray CT, acoustic emission (AE), and high-speed cameras to observe and monitor damage and manufacturing defects before, during and after testing. Figure 3 shows X-Ray CT images of the cross- and through-sections the SN005 flat panel [14], which is the basis for the analysis in this work. Several manufacturing defects can be observed in Figure 3 including misaligned tows, variability in the cross-section (X-sec) of the tows, cracking in the bulk matrix near the binder tows, and splitting cracks in the binder tows. In Figure 4, two cross-sectional images of the same C-channel specimen are displayed, and additional manufacturing defects are seen including voids, tow debonding, resin rich areas and surface cracks [14]. For the C-channel, the most severe defects are found near the web to flange transition. It is clear from Figure 3 and Figure 4, that the manufacturing-induced defects in these 3D woven composites are substantial. However, it is ambiguous as to the effect of these anomalies on the overall behavior of the component. As described previously, very little computational modeling has been conducted to understand these effects. A focus of the present work is to understand the effects of some of these manufacturing induced irregularities.

Figure 3. Pre-test X-Ray CT images of cross-section and through-section of 3D woven CFRP flat panel (SN005) exhibiting manufacturing-induced defects.
IV. Multiscale Progressive Failure Modeling of the 3D Woven Composite

A. MsRM Hierarchy of baseline SN500 3D woven model

A multiscale model of the SN500 3D orthogonal woven RUC was created for NASMAT using the weave idealized geometry, as shown in Figure 5. The bulk (inter-tow) matrix has been omitted to display the red warp, green weft, and blue binder tows. Figure 6 shows the overall MsRM RUC hierarchy deployed to model the 3D woven RUC. It can be seen that the analysis spans six length scales. Since NASMAT facilitates an integrated approach, homogenization and localization occurs across all scales at each time step to compute the macroscopic non-linear response of the RUC. Far field applied strains are passed down (localized) from the levels above, and homogenization is performed at every scale to provide the mechanical properties for the appropriate material point at the higher scale.

The MsRM framework is agnostic with respect to the specific micromechanics theory used. Hypothetically, any micromechanics theory can be used at any scale in an MsRM analysis. This allows for tailoring of the solution methodology to optimize the balance between efficiency and fidelity. Here, a mixture of the doubly-periodic GMC, triply periodic GMC, and MT were utilized, as shown in Figure 6 [25],[27].

Referring to Figure 6, Level 0, which is the highest scale, represents the composite macroscale. NASMAT computes the effective properties, as well as the fully 3D macroscopic stress and strain tensors considering all combined nonlinear effects from lower scales.

The homogenization of the mesoscale is separated into two steps, called double-homogenization [33]. At the next scale, Level 1, the mesoscale details of the 3D woven composite are represented with a single subcell through the thickness of the RUC and the effective properties from Level 2. Triply-periodic GMC is used to homogenize Level 1. At the stack-scale, Level 2, the details of the 3D weave are represented, and triply-periodic GMC is again used for the homogenization. Double-homogenization is employed to compensate for the lack of shear-normal coupling in GMC [33]. By homogenizing the stacks in Level 2 first, the effective properties in the subcells at Level 1 are anisotropic. Therefore, the first-order normal-shear coupling is retained at Level 1 as an effect of the local constitutive response, as opposed to the arrangement of constituents.

Figure 4. Pretest X-Ray CT images of cross-section of C-channel coupon specimen exhibiting manufacturing-induced defects.
Figure 5. Idealized geometry of the 3D woven SN500 RUC used for the MsRM model in NASMAT. Only the warp (red), weft (green), and binder (blue) tow representations are displayed; the bulk (inter-tow) matrix is omitted.

Figure 6. MsRM hierarchy deployed to model the SN500 3D woven CFRP within NASMAT. Level 0 – Macroscale. Level 1 – Mesoscale. Level 2 – Stack-scale. Level 3 – Microscale. Level 4 – Constituent-scale. Level 5 – Subscale.
Levels 1 and 2 are used to represent the idealized geometry presented in Figure 5. One advantage of using GMC is that the theory is insensitive to refinements in the discretization of the unit cell, for a fixed geometry. Therefore, the subcell grid with the lowest number of subcells needed to represent the desired geometry can be utilized. This is because the continuity conditions used to formulate the strain concentration tensor, which maps the global strains to the local strains, is enforced in an average, integral sense. This results in the directional volume fractions being the contributing factor in the homogenization, not local spatial location [34]. Thus, many of the challenges normally associated with meshing these complex architectures are eliminated for an idealized geometry [24]. Imperfect, or “non-ideal,” geometries can also be considered by altering the idealized RUC, as appropriate.

The effective constitutive properties of the subcells in the stacks are computed via the homogenization of the bulk matrix containing voids, using MT, and the homogenization of a 4x4 RUC of hexagonally packed fibers in the tow, using doubly-periodic GMC, at Level 3 – the microscale. Stress concentrations due to the presence of the square fiber corners in the 4x4 RUC are not captured for the same reasons that GMC is not sensitive to grid refinement. In GMC, the local fields are actually centroidal quantities, and the rectangular subcells represent the region of influence of each centroid. At Level 4 (the constituent scale), the material properties of the fiber constituent at the current time are provided to Level 3 along with the effective properties of the inter-tow matrix, obtained using MT. In addition, the effective properties of the matrix constituent, and the spherical void, are provided to the Level 3 MT model of the effective matrix containing voids. Details of how the properties of the void are treated are given in the “Material Properties” subsection below. Finally, the Level 5 subscale consists of just the base matrix and void constituents, the properties of which are used in the Level 4 MT model of the inter-tow matrix.

The baseline MsRM NASMAT model contains a total of 6,672 GMC subcells and 27,114 MT phases (voids or matrix). A total of 14,760 homogenizations are performed each time step, 13,557 of which use MT. The model has been subjected to uniaxial load cases to simulate the component-wise stress-strain responses of the material. In each case, strain-controlled loading has been simulated, wherein all stress components (other than that associated with the applied strain component) are kept at zero. For the subvolume elimination approach, the specified uniaxial strain component has been applied over 100 increments, with the average computation time for each simulation using a single CPU on a Windows machine is on the order of ~30 seconds. The crack band simulations take considerably longer as several iterations are typically required per loading increment. Particularly when the composite has lost most of its stiffness due to damage (near the end of the simulation), many iterations may be required to reach convergence. This situation could be detected to automatically end the simulation, but currently, no attempt has been made to implement this capability. Still, the crack band simulations are still quite efficient, typically taking between 10 and 40 minutes using a single CPU on a Windows machine for 200 loading increments.

**B. Constituent Failure Models**

1. **Subvolume Elimination Approach**

In the simple subvolume elimination method, material non-linearity is introduced into the model in the form of local, abrupt, material failure. Maximum stress and maximum strain failure criteria were specified for the fiber constituent that considers only the normal components of stress or strain.

\[
\text{max} \left( \left| \sigma_x^{(a, \beta, \gamma)} \right|, \left| \varepsilon_x^{(a, \beta, \gamma)} \right| \right) \geq 1, \quad \sigma_x^{(a, \beta, \gamma)} > 0, \quad i = 1, 2, 3
\]  

where \( \sigma_x^{(a, \beta, \gamma)} \), \( \varepsilon_x^{(a, \beta, \gamma)} \) are the normal components of stress and strain in the a fiber subcell, and \( X_x \), \( X_e \) are the stress and strain allowables for the fiber. For the matrix, a maximum stress criterion was also applied, but all stress components were considered,

\[
\text{max} \left( \left| \sigma_x^{(a, \beta, \gamma)} \right|, \left| \tau_x^{(a, \beta, \gamma)} \right| \right) \geq 1, \quad \sigma_x^{(a, \beta, \gamma)} > 0, \quad i = 1, 2, 3, \quad i \neq j
\]
\( \sigma_{ij}^{(\alpha, \beta, \omega_m)} \) and \( \tau_{ij}^{(\alpha, \beta, \omega_m)} \) are the normal and shear stresses in a matrix subcell (or phase for MT), and \( X_m \) and \( Y_m \) are the normal and shear stress allowables for the matrix. Upon satisfaction of Eq. (12) or (13), the simple subvolume elimination damage model sets the stiffness of the failed subcell (or phase for MT) to a very low value (in all directions). That is, the damage progression is a step function, with complete subcell failure (or phase failure for MT) occurring instantaneously. Only the fiber and matrix constituent materials, which are at the lowest scales (Levels 5 and 6) in the MsRM hierarchy (see Figure 6) are failed in this methodology. Yet, even though these failures are discrete, the effective damaging response at the macroscale is progressive as the subcell/phase failures accumulate locally, and the nonlinear effects of failure percolate up the scales through the homogenized stiffness tensors.

2. Crack Band Model

In contrast to the subvolume elimination approach, the crack band model is, in general, gradual at the scale at which it is applied. Herein, like subvolume elimination, it has been applied at the lowest, constituent, length scale. However, now the damage progression can evolve as opposed to being abrupt.

The crack band model, originally proposed in ref. [29], has been modified and implemented within NASMAT to model failure in the matrix phase, both bulk (inter-tow) and within the tow (intra-tow), of the 3D woven composite. This continuum damage theory assumes that the fracture energy due to cracking in the material is dissipated over the entire volume of the continuum element containing the damaging zone, or crack band. The relationship between the traction on and separation of the crack faces within the crack band are governed by a traction versus separation law, and the area under the traction-separation law is equal to the fracture toughness of the material. The original formulation assumed that the crack band was oriented such that the crack faces were oriented normal to the direction of maximum principle stress. This formulation was previously utilized to model cracking within the GMC and HFGMC micromechanics theories using NASA’s MAC/GMC code [30]. In the current model, it is assumed that the crack band aligns normal to one of the basis vectors in the local material coordinate frame, and the cracks evolve under a mixed-mode relationship, see Figure 7.

![Figure 7. Crack band model damage modes.](image)

The development of the mixed-mode traction separation law for the crack band model follows that used in a popular cohesive zone element formulation [31]. Initiation of the crack band is governed by a set of quadratic failure criteria,

\[
F_1 = \left( \frac{\sigma_1^{(\alpha, \beta, \omega_m)}}{X_m} \right)^2 + \left( \frac{\tau_{12}^{(\alpha, \beta, \omega_m)}}{Y_m} \right)^2 + \left( \frac{\tau_{13}^{(\alpha, \beta, \omega_m)}}{Y_m} \right)^2
\]

\[
F_2 = \left( \frac{\sigma_2^{(\alpha, \beta, \omega_m)}}{X_m} \right)^2 + \left( \frac{\tau_{12}^{(\alpha, \beta, \omega_m)}}{Y_m} \right)^2 + \left( \frac{\tau_{23}^{(\alpha, \beta, \omega_m)}}{Y_m} \right)^2
\] (14)
\[ F_3 = \left( \frac{\sigma_3 (\alpha m \beta m r_m)}{X_m} \right)^2 + \left( \frac{\tau_{13} (\alpha m \beta m r_m)}{Y_m} \right)^2 + \left( \frac{\tau_{23} (\alpha m \beta m r_m)}{Y_m} \right)^2 \]

where \( F_i \) are failure indices. \( X_m \) and \( Y_m \) are the normal and shear matrix cohesive strengths, which govern damage initiation. It should be noted that, the strengths in Eq. (14) are cohesive strengths which are numerical parameters and are not necessarily related to the strengths obtained from tension or torsion experiments, used in Eq. (13). Once one of the failure indices exceeds unity, the orientation of the crack band and tractions are determined.

\[ F_1 \geq 1; \ i = 1, j = 2, k = 3 \]

\[ F_2 \geq 1; \ i = 2, j = 3, k = 1 \]

\[ F_3 \geq 1; \ i = 3, j = 1, k = 2 \]  \hspace{1cm} (15)

where \( i, j, \) and \( k \) are the indices of the material coordinates in Figure 7, and the crack band tractions \( (t_I, t_{II}, \) and \( t_{III}) \) are related to the subcell stresses.

\[ t_I = \sigma_i (\alpha m \beta m r_m); \ t_{II} = \tau_{ij} (\alpha m \beta m r_m); \ t_{III} = \tau_{kl} (\alpha m \beta m r_m) \]

(16)

Since damage growth in the crack band affects the constitutive response of a continuum, a mixed-mode traction, \( t_M \), is defined and related to an equivalent mixed-mode strain, \( \varepsilon_M \), through a mixed-mode traction-strain law depicted in Figure 8. Here, a triangular traction-strain law is employed, but various shapes can be utilized. Assuming a triangular law,

\[ t_M = (1 - d_M)E_M \varepsilon_M \]  \hspace{1cm} (17)

\[ d_M = \left\{ \begin{array}{ll}
0; & \varepsilon_M \leq \varepsilon_M^0 \\
1 - D_M; & \varepsilon_M^0 < \varepsilon_M < \varepsilon_M^f \\
1; & \varepsilon_M \geq \varepsilon_M^f \end{array} \right. \]  \hspace{1cm} (18)

\[ D_M = \frac{\varepsilon_M^0 (\varepsilon_M^f - \varepsilon_M)}{\varepsilon_M (\varepsilon_M^f - \varepsilon_M^0)} \]  \hspace{1cm} (19)

where \( \varepsilon_M^0 \) is the mixed-mode strain when damage initiates according to Eqs. (14) and (15), and \( \varepsilon_M^f \) is the mixed-mode strain at the end of the traction-strain law in Figure 8. The mixed-mode initiation strain can be determined from the mixed-mode initiation criterion given by Eqs. (15) and (16),

\[ \varepsilon_M^0 = \varepsilon_{II}^0 \sqrt{\frac{1 + \beta^2}{(\varepsilon_{II}^0)^2 + (\beta \varepsilon_{II}^0)^2}} \]  \hspace{1cm} (20)

and \( \beta \) is the mode-mixity parameter,
\[
\beta^2 = \frac{\gamma_{II}^2 + \gamma_{III}^2}{\varepsilon_l^2} \tag{21}
\]

The individual strains, \(\varepsilon_l, \gamma_{II}, \gamma_{III}\), associated with the fracture modes are,

\[
\varepsilon_l = \varepsilon_l^{(\alpha_m \beta_m \gamma_m)}, \quad \gamma_{II} = \gamma_{II}^{(\alpha_m \beta_m \gamma_m)}, \quad \gamma_{III} = \gamma_{III}^{(\alpha_m \beta_m \gamma_m)} \tag{22}
\]

and the initiation strains, \(\varepsilon_l^0, \gamma_{II}^0\), associated with the fracture modes are related to the cohesive strengths, the undamaged Young’s modulus \(E\) and shear modulus \(G\), assuming the material is isotropic,

\[
\varepsilon_l^0 = \frac{X_m}{E}; \quad \gamma_{II}^0 = \frac{Y_m}{G} \tag{23}
\]

The final mixed-mode strain, \(\varepsilon_M^f\), is calculated from the mixed-mode fracture criterion. Here, a mixed-mode power law is assumed,

\[
\left( \frac{G_l}{G_{IC}} \right)^\alpha + \left( \frac{G_{II}}{G_{IIIC}} \right)^\alpha = 1 \tag{24}
\]

where \(G_{IC}\) and \(G_{IIIC}\) are the mode I and mode II fracture toughness of the material, respectively. The behavior of microcracks under mode III conditions is still an active area of research. For simplification, it is assumed in Eq. (24) that the matrix has no fracture resistance under mode III cracking. The mode I and mode II strain energy release rates \(G_l\) and \(G_{II}\) are calculated as

\[
G_l = l_c \int t_l d\varepsilon_l \tag{25}
\]

\[
G_{II} = l_c \int t_{II} d\gamma_{II} \tag{26}
\]
Assuming triangular traction-separation laws for the individual modes, similar to the mixed-mode triangular traction-strain law in Figure 8, and that the mode-mixity, $\beta$, remains constant in the subcell throughout the evolution of the crack band, the strain energy release rates upon satisfaction of Eq. (24) can be calculated.

$$G_I = \frac{l_c X_m \varepsilon_m^f}{2\sqrt{1 + \beta}} \quad (27)$$

$$G_{II} = \frac{1}{2} l_c Y_m \varepsilon_m^f \sqrt{1 + \beta} \quad (28)$$

The strain energy release rates are related to the strain energy density through a characteristic length $l_c$ [29]. In the context of the GMC and HFGMC micromechanics models, this length is the dimension of the subcell, see Figure 1, normal to the crack band.

$$l_c = \left\{ \begin{array}{ll} d_{\alpha m}; & i = 1 \\ h_{\beta m}; & i = 2 \\ l_{\gamma m}; & i = 3 \end{array} \right. \quad (29)$$

Utilizing Eqs. (27) and (28) in Eq. (26), $\varepsilon_m^f$ can be calculated.

$$\varepsilon_m^f = \frac{2(1 + \beta^2)}{l_c \varepsilon_m^0} \left[ \left( \frac{E}{G_{IC}} \right)^{\alpha} + \left( \frac{\beta^2 G}{G_{II}} \right)^{\alpha} \right]^{-1/\alpha} \quad (30)$$

Eq. (30) dictates that the traction-strain law presented in Figure 8 is dependent on the local dimensions of the subcell containing the crack band. Therefore, the damage evolution law in Eq. (19) is also a function of the characteristic length.

Since the softening induced by the crack band affects the constitutive response of a continuum, the appropriate constitutive relationship must be degraded accordingly. In this case the degraded components of the compliance matrix $S^{(\alpha m_{\beta m_{\gamma m}})}$, where $\varepsilon^{(\alpha m_{\beta m_{\gamma m}})} = S^{(\alpha m_{\beta m_{\gamma m}})} \sigma^{(\alpha m_{\beta m_{\gamma m}})}$, are given as,

$$S^{(\alpha m_{\beta m_{\gamma m}})}_{li} = \frac{\varepsilon_i^{(\alpha m_{\beta m_{\gamma m}})} - \frac{u}{E} (\sigma_j^{(\alpha m_{\beta m_{\gamma m}})} + \sigma_k^{(\alpha m_{\beta m_{\gamma m}})})}{EDM \varepsilon_i^{(\alpha m_{\beta m_{\gamma m}})}}$$

$$S^{(\alpha m_{\beta m_{\gamma m}})}_{lj} = \frac{G}{D_M}$$

$$S^{(\alpha m_{\beta m_{\gamma m}})}_{jk} = \frac{G}{D_M} \quad (31)$$

At a given applied loading level (i.e., load increment), if the compliance is changed at the local constituent level, the local and global stress and strain fields will no longer be in alignment unless the global fields are recalculated based on the new local compliances. Within NASMAT, the error between the applied global stress and the volume averaged local stresses can be calculated as,

$$Err = \frac{\| \bar{\sigma} - \frac{\bar{\varepsilon}}{\varepsilon_{\alpha}} \sigma^{(\alpha)} \|}{\| \bar{\sigma} \|} \quad (32)$$
If Eq. (15) is satisfied for a given subcell, the degraded properties for those subcells are calculated with Eq. (31). The solution procedure, containing homogenization and localization steps, iterates until the value of Eq. (32) is below some threshold, indicating that the local and global fields are in alignment. Then, initiation of the crack band in additional subcells is checked using Eq. (15). If additional crack bands are initiated, the iteration procedure repeats. Thus, convergence is not achieved until both the local stresses and the state of the materials (damage/undamaged) has fully converged. In effect, this iteration and convergence procedure guarantees that, locally, every point that has initiated damage is at its correct point along the softening portion of traction-strain law shown in Figure 8.

C. Geometry Details and Material Properties

The 3D woven SN500 composite panel was composed of AS4-6k carbon fiber tows and RTM6 epoxy matrix. The minimum number of elastic constants used to compute the stiffness tensor of these phases is given in Table 1, along with the employed material strengths and matrix toughness values. The fiber behavior has always been modeled using the subvolume elimination approach, whereas results using both damage models for the matrix are presented below. The stiffness of voids was assigned an extremely low value.

<table>
<thead>
<tr>
<th>Table 1. Elastic properties of AS4 fiber and RTM6 matrix.</th>
</tr>
</thead>
<tbody>
<tr>
<td>AS4 Fiber Properties</td>
</tr>
<tr>
<td>$E_{11}$ (GPa)</td>
</tr>
<tr>
<td>$E_{22}$ (GPa)</td>
</tr>
<tr>
<td>$v_{23}$</td>
</tr>
<tr>
<td>$v_{12}$</td>
</tr>
<tr>
<td>$G_{12}$ (GPa)</td>
</tr>
<tr>
<td>$X_f$ (GPa)</td>
</tr>
<tr>
<td>$X_f^\gamma$</td>
</tr>
</tbody>
</table>

Figure 9 shows the matrix normal (mode I) and tangential (mode II) traction-separation behavior for the subvolume elimination approach and the crack band model. Note that traction-separation behavior and traction-strain behavior (see Figure 8) are related via the characteristic length (separation equals strain times characteristic length). Results representing the cases where the crack band model damage initiation strengths (see Table 1) have been decreased to 75% and 50% of their original values are also shown. The subvolume elimination approach is (locally) brittle; as soon as the damage initiates, the stiffness of the subvolume is instantaneously set to a very low value, which results in the complete loss of the subvolume’s ability to support any stress (in all directions). In contrast, the crack band model gradually reduces the appropriate stiffness components such that the traction-separation behavior is linear after damage initiates (in the present implementation). As shown, the present implementation of the crack band model adjusts the post-initiation slope such that the area under the curve is preserved, such that, when combined with the characteristic length, the correct toughness values from Table 1 are reproduced. That is, even though the matrix damage initiation stress (traction) is reduced, the mode-specific toughness of the matrix is always maintained.

The woven composite RUC geometric dimensions were taken as averages from X-ray CT image measurements. These dimensions are summarized in Table 2. The local fiber volume fraction used in the GMC RUC tows at Level 3 (see Figure 6) is 80% which results in a global fiber volume fraction of 50.3%. X-Ray CT images of the flat panels in Figure 3 do not indicate that voids are relevant manufacturing-induced defects. However, the results from acid digestions suggest that the void content is 0.4%, and therefore must be in the form of distributed voids or porosity. The local void content of the MT models is prescribed to be 0.805% which yields a global 0.4% distributed void content in the 3D woven RUC. Due to the idealization of the geometry, the global fiber volume fraction does not exactly match the fiber volume content measure from the acid digestion tests; however the global void content is accurate [12].

The subvolume elimination and crack band damage models operate at two scales, that of the inter-tow matrix and that of the matrix within the tows, as shown in Figure 6. As discussed in Section IV.B, while the subvolume elimination method is independent of the length scale, the crack band model depends on a characteristic length. Herein, for the inter-tow matrix, the crack band characteristic length has been taken from the subcell dimensions in
Table 2, which accurately represent the 3D woven composite. The implementation within NASMAT utilizes the appropriate subcell dimension given the mode of the damage. In contrast, for the matrix within the tows, the characteristic length is based on the fiber diameter given in Table 2. The subcell dimensions within the tows are set such that this fiber diameter is correctly represented, and then the matrix subcells correctly represent the distances between the fibers.

Figure 9. Matrix (a) normal (mode I) and (b) tangential (mode II) traction-separation behavior for the subvolume elimination approach and the crack band model. Results representing the cases where the crack band model damage initiation strengths have been decreased to 75% and 50% are also shown.

<table>
<thead>
<tr>
<th>Table 2. RUC geometric dimensions.</th>
</tr>
</thead>
<tbody>
<tr>
<td><strong>Binder Tows</strong></td>
</tr>
<tr>
<td><strong>Width (mm)</strong></td>
</tr>
<tr>
<td><strong>Warp Tows</strong></td>
</tr>
<tr>
<td><strong>Width (mm)</strong></td>
</tr>
<tr>
<td><strong>Distance Between (mm)</strong></td>
</tr>
<tr>
<td><strong>Weft Tows</strong></td>
</tr>
<tr>
<td><strong>Width (mm)</strong></td>
</tr>
<tr>
<td><strong>Distance Between (mm)</strong></td>
</tr>
<tr>
<td><strong>Fiber Diameter (µm)</strong></td>
</tr>
</tbody>
</table>

D. Inclusion of Matrix Voids and Binder Tow Disbonding

As indicated in Figure 6, distributed voids in both the inter-tow and intra-tow matrix have been included in the MsRM model of the 3D woven composite using the MT model. As mentioned, the local void content of the MT models has been prescribed to be 0.805% to match the measured global 0.4% distributed void content.

As discussed in Section III, disbonding of the binder tows in the warp direction as they traverse the composite in the through-thickness (TT) direction, is a prominent observed feature in the 3D woven AS4/RTM6 composite. Typical binder tow disbonds present in the as-manufactured AS4/RTM6 composite are shown in Figure 3. To model the effect of such preexisting disbonds on the mechanical response of the composite, a different GMC representation of the through-thickness section of the binder tows has been defined. That is, the binder tows running in the warp direction, at the top and bottom of the composite, have not been redefined; only the TT portions of the binder tow are altered. The new TT binder tow definition includes a thin subcell (adjacent in the x_3-direction) that traverses the entire thickness with very low mechanical properties. This disabled the ability of the TT binder tow to transfer any traction in the x_3-direction (warp), see Figure 5. This has been applied to all six TT binder tows and thus represents an upper bound for the effect of the TT binder tow disbonds on the composite response. In the context of the MsRM model of
the composite, the thin void is added adjacent to the homogenized through-thickness stack of subcells representing
the TT binder tow, essentially forming a new level between the mesoscale (Level 1) and the stack-scale (Level 2), at
the locations where there are TT binder tows, shown in Figure 5 and Figure 6.

V. Results and Discussion

The predicted warp-direction ($x_3$, see Figure 5) stress-strain curves, when using the subvolume elimination
approach and the crack band model, are compared to experimental data in Figure 10. Good agreement between the
simulated and experimentally observed stress-strain curves (using an ARAMIS digital image correlation system) were
obtained with minimal analysis assumptions. Comparisons of the predicted warp-direction Young’s modulus and
ultimate tensile strength (UTS) are made in Table 3. The model predictions are slightly stiffer than the experimental
measurements, while the predicted UTS values using the subvolume elimination approach and the crack band model
are within the test results (which do exhibit some scatter). Comparing the results of the two damage models, the
primary distinctions in the noticeable drop in the subvolume elimination stress-strain curve at 343 MPa due to failure
in bulk matrix. This is caused by the brittle nature of the matrix in subvolume approach vs. the (generally) more
gradual failure in the crack band model (see Figure 9). As a result, the crack band model prediction follows the test
data much better than the subvolume elimination approach. Both damage models predict the average ultimate failure
within 5.3% on average. Note that, if the effects of the fiber strength statistics were included, the crack band UTS and
strain to failure predictions would likely be reduced toward the test results plotted in Figure 10.

![Figure 10. Warp-direction stress-strain predictions using the subvolume elimination approach and the crack band model compared to experimental data.](image)
Table 3. Comparison of warp direction test results and model predictions.

<table>
<thead>
<tr>
<th>Result</th>
<th>Warp Young’s Modulus (GPa)</th>
<th>Warp Ultimate Stress (MPa)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Test 1</td>
<td>55.8</td>
<td>798</td>
</tr>
<tr>
<td>Test 2</td>
<td>55.5</td>
<td>763</td>
</tr>
<tr>
<td>Test 3</td>
<td>54.7</td>
<td>815</td>
</tr>
<tr>
<td>Test 4</td>
<td>59.8</td>
<td>761</td>
</tr>
<tr>
<td>Test 5</td>
<td>62.4</td>
<td>710</td>
</tr>
<tr>
<td>Test Average</td>
<td>57.7</td>
<td>769</td>
</tr>
<tr>
<td>Test Standard Deviation</td>
<td>3.3</td>
<td>% Error from avg. 41</td>
</tr>
<tr>
<td>Subvolume Elimination</td>
<td>59.3</td>
<td>2.9</td>
</tr>
<tr>
<td>Crack Band</td>
<td>59.3</td>
<td>810</td>
</tr>
<tr>
<td></td>
<td></td>
<td>% Error from avg. 5.3</td>
</tr>
</tbody>
</table>

Six uniaxial, applied strain simulations were conducted to obtain the six component-wise nonlinear stress-strain curve predictions for the 3D woven composite shown in Figure 11. In all cases, the specified strain component was applied, while all other stress components were kept at zero, thus simulating a strain-controlled uniaxial test. The subvolume elimination approach has been compared to the crack band approach wherein the matrix damage initiation stresses ($X_m$ and $Y_m$) has been varied to consider cases where these stresses are reduced to 75% and 50% of the original values in Table 1. As discussed in Section IV.C, when the failure initiation stress is altered, the crack band model implementation automatically adjusts the unloading slope in the traction-separation behavior (see Figure 9) such that the toughness values (given in Table 1) are maintained.

As shown in Figure 11, the impact of the damage model and crack band damage initiation stress is reduced when the 3D woven composite is more fiber dominated. The weft-direction ($x_2$) normal loading (Figure 11b) represents the most fiber dominated case (as tows are continuous in this direction), and the damage has minimal effect prior to predicted failure via fiber failure. The warp-direction ($x_3$) behavior (Figure 11c) is also quite fiber dominated, but because of the presence of the binder tows, which run in the $x_3$-direction (see Figure 5), the damage model and crack band initiation stress have a noticeable effect. As shown, lowering the crack band initiation stress, which allows matrix damage to initiate earlier, lowers the predicted stress strain curve, but does not significantly impact the predicted strain to failure.

The remaining loading orientations (Figure 11a,d,e,f) are not fiber dominated, and thus the damage model and crack band initiation stresses have a significant impact. In the through-thickness direction ($x_1$), the crack band model predictions are much more gradual compared to the subvolume elimination result (see Figure 11a), which has a zig-zag shape as discrete brittle matrix failures occur in the composite. Among the crack band predictions, the damage initiation stress has a significant effect as the earlier matrix damage significantly lowers the predicted response. These crack band model stress-strain simulations were terminated at an applied strain of 0.03, but it appears that all three had reached their ultimate stress by this point.

In the in-plane shear stress-strain curves (see Figure 11a), the crack band predictions, like the subvolume elimination predictions, exhibit a zig-zag shape, despite the matrix damage behavior being much more gradual. It should be noted that this more gradual behavior is local, thus, at a given global loading level, it is possible that the local matrix behavior can progress from the initiation point all the way to the zero traction separation shown in Figure 9. Lowering the crack band damage initiation stress lowers the deviation from linearity in the in-plane shear stress-strain curves, as one would expect. Interestingly, however, the predicted global damage progression is quite different among the three crack band cases. The axial shear predictions (see Figure 11e and f) also show the lower deviation from linearity associated with the lower crack band damage initiation. These responses, however, are more brittle compared to the in-plane shear response, so the impact of the damage model on the global damage progression is less pronounced.

Figure 12 compares stress-strain predictions for the subvolume elimination approach and crack band model with and without a preexisting disbond of the binder tows. Because the binder tow disbond is normal to the $x_3$-direction (see Figure 3 and Figure 4), its effect on the through-thickness ($x_1$) direction and the weft ($x_2$) direction normal stress-strain responses is minimal (see Figure 12a and Figure 12b). As one would expect, the effect in the warp ($x_3$)-direction (see Figure 12c) is more noticeable. The initial warp-direction Young’s modulus is decreased by 8%, but after the damage initiations, the crack band curve and the two subvolume elimination curves match quite closely.
Figure 11. Predicted stress-strain responses of the 3D woven composite using the subvolume elimination approach and the crack band model with varying damage initiation stresses. (a) Applied $\varepsilon_{11}$ (through-thickness) strain. (b) Applied $\varepsilon_{22}$ (weft) strain. (c) Applied $\varepsilon_{33}$ (warp) strain. (d) Applied $\gamma_{23}$ (in-plane shear) strain. (e) Applied $\gamma_{13}$ (transverse shear). (f) Applied $\gamma_{12}$ (transverse shear).
Figure 12. Predicted stress-strain responses of the 3D woven composite, with binder tow disbonds included, using the subvolume elimination approach and the crack band model. (a) Applied $\varepsilon_{11}$ (through-thickness) strain. (b) Applied $\varepsilon_{22}$ (weft) strain. (c) Applied $\varepsilon_{33}$ (warp) strain. (d) Applied $\gamma_{13}$ (transverse shear). (e) Applied $\gamma_{12}$ (transverse shear).
Note that the in-plane shear response is not plotted as adding the binder tow disbond through the entire thickness of the 3D woven composite completely eliminates the in-plane ($x_2$-$x_3$) shear stiffness. As previously discussed, this is because of the well-known lack of shear coupling within the GMC theory being used at Level 1 (see Figure 6) in the employed MsRM approach. This is obviously unrealistic and can be overcome by reducing the extent of the disbond or employing the HFGMC micromechanics theory [27].

The transverse shear response curves are shown in Figure 12d and e. While the $\tau_{12}$-$\gamma_{12}$ response (Figure 12e) is insensitive to the presence of the binder tow disbond, the $\tau_{13}$-$\gamma_{13}$ response (Figure 12d) is significantly impacted. The initial shear modulus is reduced by 28%, and the ultimate stress is reduced by 19%. This is again due to the orientation of the disbond, normal to the $x_3$-direction, and the lack of normal shear coupling may be a contributor to the severity. It is noteworthy, however, that such an apparently severe preexisting feature such as binder tow disbonds (see Figure 3) is predicted to have a relatively minor overall impact on the composite nonlinear failure behavior (as shown in Figure 12). This is likely part of the 3D woven composite’s desirable damage tolerance characteristics.

VI. Conclusion

The MsRM method, utilizing the subvolume elimination approach and the crack band model, was deployed within the NASMAT computational framework to predict the stress-strain progressive failure response of a 3D woven composite unit cell containing as manufactured distributed voids and pre-existing binder tow disbonds. The warp-direction Young’s modulus and UTS were predicted within 5.3% of experimental data, with the two damage model predictions within the scatter in the experimental UTS values. The simulation results provide insight into the damage tolerant nature of 3D woven composites. Although initial matrix damage occurs and can be observed in the global stress-strain response of the composite, the cracks are not allowed to progress and develop into critical flaws. Catastrophic failure does not occur until much later, as a result of fiber fracture. In addition, the in-plane damage tolerance of the 3D woven material was demonstrated because of its ability to carry substantial load in the presence of porosity and binder tow disbonds.

Parametric studies were conducted to ascertain the influence of the matrix damage initiation stress employed in the crack band model on the stress-strain responses under different applied strain components. The crack band implementation preserves the area under the traction-separation curves when this damage initiation stress is altered, resulting in the material toughness being maintained (when combined with the characteristic length). The effect of the damage initiation stress was shown to be muted in fiber-dominated loading situations (warp and weft normal loading), while very impactful in the other loading situations. The effect of disbonding of the binder tows was also simulated using both the subvolume elimination approach and the crack band model. Here, the greatest impact was on one of the transverse shear responses, with little effect observed for the other loading components.

The predictive capability of NASMAT combined with its speed provides an attractive tool for performing rapid engineering trade studies on complex composite systems. It was shown that the simpler subvolume elimination is much more computationally efficient than the more physical crack band model as the latter requires iterations at each loading increment. As such, a trade off between computational efficiency and fidelity can be made when selecting the damage modeling approach for a given application. The application to the 3D woven composite considered herein also demonstrates NASMAT’s flexibility by incorporating six separate levels of calculations and three distinct micromechanics theories under one platform. This development addresses an apparent technological gap observed in the open literature. There exist very few modeling tools that can delve below the meso (homogenized tow) scale of 3D woven composites even though this may be necessary to capture the physics of damage at the appropriate length scale.

Acknowledgments

Primary NASMAT development has been supported through the NASA Aeronautics Research Mission Directorate’s (ARMD’s) Transformational Tools and Technologies (TTT) Project. Additional support, specific to MsRM, came from the Space Technology Mission Directorate’s Composite Technology for Exploration (CTE) Project and ARMD’s Advanced Composites (AC) Project.
References


