Multi scale modeling and characterization of inelastic deformation mechanisms in continuous fiber and 2D woven fabric reinforced metal matrix composites

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Abstract

Continuous unidirectional ceramic fiber and woven fabric reinforced metal matrix composites (MMCs) have potential to obtain very high specific strengths and stiffnesses, but use in structural applications has thus far been limited by their inherently low ductility, particularly in tensile loading conditions. In this work, a multi-scale micromechanics based finite element framework is used to predict, and understand the effect of microstructure on the tensile deformation behavior, including progressive damage and failure, of ceramic fiber and fabric reinforced MMCs. A hierarchal approach is implemented in which a micro-scale model is used to determine the transversely isotropic elasto-plastic mechanical behavior of unidirectional fiber reinforced MMC based on the properties of an aluminum alloy matrix, individual ceramic fibers, and the fiber-matrix interface. The validated transversely isotropic constitutive behavior is then input into a unit cell model for a woven fabric MMC consisting of unidirectional MMC tows in

an aluminum matrix. Parallel experimental tensile testing and characterization of fracture mechanisms are used to validate our model for unidirectional and 2D weave fabric MMC. Cohesive zones are used to model the interfacial properties at both scales and we are able to quantify the contribution of various deformation and damage mechanisms such as ductile matrix failure, interfacial decohesion, transverse plasticity, and axial fiber fracture to the overall mechanical response of the MMC. It was found that interfacial debonding contributes significantly to the inelastic response of unidirectional and woven fabric MMC and that experimental data are relatively well reproduced by a model that assumes a weak interfacial debonding cohesive law (maximum allowable interface stresses on the order of 100 MPa and total work of decohesion = 75 J/m²).

Keywords: metal matrix composites (MMCs); mechanical properties; finite element method; Micromechanics; cohesive zone modeling

1. Introduction

High performance continuous fiber reinforced metal matrix composites (MMC) have long been researched as potential alternatives to traditional structural alloys such as steel and aluminum [1]. A great deal of the attractiveness of ceramic fiber MMCs comes from the fact that they can potentially achieve very high specific strength and stiffness to provide higher performance and mass efficient alternatives to their monolithic metallic counterparts [2]. Additionally, these materials give the design engineer a relatively wide design space as the strength and stiffness of a particular composite can be fine tuned by varying reinforcement

morphology, volume fraction and alloy characteristics to meet to the specifications of a particular application [2-8]. Unidirectional fiber reinforced MMCs offer the greatest increases in strength and stiffness when loaded in along the axis of the fibers, however the transverse properties are often very poor. Reinforcing the transverse direction by using cross ply laminates or by incorporating fabric structures increases the transverse strength but reduces the tensile ductility of the material. As a consequence, the use of continuous ceramic fiber reinforced MMCs in structural applications have been fairly limited due to the overall poor tensile ductility and brittle failure characteristics thus far observed in these materials.

Due to the large number of material design variables possible in MMC systems, modeling becomes a de facto aspect of the material design workflow. Analytical methods such as Eshelby's inclusion theory [9] and the Mori-Tanaka method [10], and numerical models of composites are commonly used tools to predict the mechanical response and understand the underlying deformation behavior of metal matrix composite systems [2]. Numerical micromechanical based approaches using the finite element method have been used to study a wide range of elastic and inelastic metal matrix composite deformation behaviors. Finite element methods have been applied to study the major deformation and failure mechanisms of fiber MMCs to include matrix plasticity [11-13], interface debonding [14-16], and fiber fracture [17-18]. The vast majority of the literature pertaining to metal matrix composites is aimed at understanding the mechanics of unidirectional composites or cross-ply laminates with comparatively little literature explicitly addressing woven fabric MMC. Although not specific to MMC, numerical finite element methods, including multi-scale approaches, have been successfully used to study the mechanics and failure mechanisms of fabric reinforced polymer matrix composites including non-linear effects of damage and failure [19-22].

Most of the experimental characterization of the mechanical behavior of continuous unidirectional alumina fiber reinforced aluminum that has been reported in the literature has focused on studies of the strength and deformation mechanisms of the composite loaded along the fiber axis. Compressive [23-24], tensile [25-28], bending [25], and fatigue [29] properties have been reported for loading applied along the fiber axis of unidirectional alumina fiber reinforced aluminum. Hall et al [30] studied the quasi-static and high strain rate properties of cross-ply unidirectional alumina reinforced aluminum MMC and found that the properties were strongly dependent on fiber distribution and alignment. Little experimental literature has been reported on the transverse properties of unidirectional alumina fiber reinforced aluminum or on the mechanical behavior of 2D woven fabric reinforced metal matrix composites.

In this paper a multi-scale finite element framework is developed to predict the transversely isotropic tensile deformation response, including progressive damage and failure, of continuous unidirectional fiber and woven fabric metal matrix composites based on the isotropic properties of the constituent ceramic fiber and aluminum alloy matrix. A hierarchical multi-scale approach is implemented in which the predicted transversely isotropic constitutive behavior of the unidirectional composite is used as an input for MMC tows in a unit cell model of a 2D weave MMC to predict homogenized properties at the macroscale. Experimental characterization of both materials is used to validate the numerical results and to characterize the active deformation and failure mechanisms occurring in each composite system. The validated models are then used to quantify the relative contributions of individual damage mechanisms such as transverse plasticity and decohesion of the fiber-matrix interface to the overall strain energy dissipated by the composite.

2. Experimental

2.1 Materials

Unidirectional continuous alumina (Al_2O_3) fiber reinforced MMC specimens were produced using an inert gas pressure infiltration casting process by CPS Technologies (Norton, MA, USA). The ceramic fiber was Nextel 610TM (3M, St. Paul, MN, USA) having a denier of 10,000 and was provided in laminated performs. The fiber is greater than 99% α - Al_2O_3 with a diameter of 10-12 μm and density of 3.9 g/cm^3 [31]. A second set of MMC specimens were made using the same manufacturing process using a 2D plain weave fabric produced from Nextel 610TM yarns. The Nextel yarns produced from the 10,000 denier fiber have a nominal filament count per yarn of 2550 [31]. The matrix alloy used was Al-2% Cu (0.01 wt% Si, 0.01 wt% Fe and 2.17 wt% Cu). A casting of the matrix alloy was also produced for characterization using the same processing parameters and thermal history as the fiber reinforced composites. The final volume fraction of fiber was measured using image analysis of the micrographs of the polished surfaces and found to be 55% in the unidirectional composite and 30-35% in the fabric MMC. The variability is due to non-uniform compression of the fabric layers during infiltration with the molten aluminum. Microstructures of polished unidirectional and fabric reinforced MMC materials are shown in Figures 1 and 2 respectively. The unidirectional microstructure is taken transverse to the axial direction of the fibers. The microstructures of the fabric MMC are taken normal to the plane of the weave. In the lower magnification micrograph the warp and weft yarns can be seen, while in the higher magnification micrograph the fibers in the sub-yarn structure can be seen to be in a very similar packing arrangement and density as the unidirectional MMC. In both materials, regions of porosity (black) can be observed particularly at fiber-matrix interfaces. Formation of micro-porosity in this case is likely due to micro shrinkage which occurs when liquid metal cannot reach interdendritic areas during solidification

and is a standard foundry defect for mushy freezing alloys. Additionally, the intermetallic eutectic θ phase Al_2Cu (white) can be observed dispersed along the fiber-matrix interfaces. Intermetallic phases along fiber-matrix interfaces are common microstructural features/defects that develop during solidification of Al-Cu alloys during the casting process [32]. From the Al-Cu phase diagram, the solid solution phase (α) is the first to solidify while most of the copper remains in liquid form. As solidification progresses intermetallic particles of the CuAl_2 (θ) phase form in the solid state region below the solvus line. As the region is near fibers are often the last to solidify it is common to find the precipitates around the reinforcement phases in MMC. Due to the large mismatch of coefficients of thermal expansion (CTE) between the ceramic fiber and the Al-2%Cu matrix, residual stresses large enough to cause strain hardening in the matrix are expected [33] as the composite cools during the casting process. The CTE of the Nextel fibers is $8 \times 10^{-6} \text{ K}^{-1}$ [31], and the CTE of the matrix is $\sim 20 \times 10^{-6} \text{ K}^{-1}$. The resulting strain hardening due to this mismatch will alter the in situ properties of the matrix in the composite compared to the as-cast unreinforced matrix [27, 33-35] and needs to be accounted for in the modeling to accurately predict the properties and response of the composite. To obtain a semi-quantitative estimate of the in-situ plastic flow properties of the Al-2%Cu matrix, nano-hardness measurements were conducted in the as-cast unreinforced alloy and in the as-cast MMC material. This method gives an approximate, phase averaged, matrix hardness and does not account for local variations in hardness due to inhomogeneous matrix strains, particularly in the close vicinity of particles where the density of geometrically necessary dislocations (GND) are expected to be highest [36-37]. This method has shown previous success in determining the in situ matrix properties of unidirectional metal matrix composites [38]. The MTS NanoIndenter XP equipped with the Test Works 4 software was used. Polished samples were indented in the

matrix phase in between fibers using the continuous stiffness method with a Berkovich indenter. Based on the average of ten measurements it was found that the in situ hardness of the matrix in the unidirectional MMC was 1.28 times harder than the as cast Al-2%Cu alloy. The original experimental and scaled uniaxial tension stress strain curves for the Al-2%Cu are presented in Figure 3.

2.2 Quasi-static tensile testing

Quasi-static tensile testing was conducted to determine constitutive properties of the Nextel fibers and Al-2%Cu as-cast matrix and laminate properties of the unidirectional MMC and woven MMC in the as-cast condition. Single fiber tension tests measured the ultimate tensile strength and failure strain properties of the Nextel 610TM fiber to be 2970 ± 730 MPa and 0.77% respectively [39]. These values are similar to values previously reported by Wilson [40-41]. Uniaxial tension testing was conducted for the Al-2%Cu alloy as well as the unidirectional and fabric reinforced composite materials using flat rectangular specimens with a gage length of 89 mm. Uniaxial tension testing was conducted according to ASTM D3552 (standard test method for tensile properties of fiber reinforced MMC) on an MTS test frame at a strain rate of 0.001 s^{-1} . Alignment of the load frame grips was achieved within the ASTM E1012 specification. Aluminum sheeting with a 0.51 mm thickness was wrapped around the grip sections of each sample to help prevent grip failures from debris or the rough grip surfaces. The unidirectional MMC was tested in two fiber orientations with respect to the machine loading axis: 0° (axial) and 90° (transverse). The fabric reinforced MMC was tested within the plane of the fabric with the warp fiber direction aligned with the loading axis of the machine.

3. Numerical modeling of unidirectional fiber reinforced MMC

3.1 Model setup

A 3D unit cell finite element model (FEM) was used to simulate the deformation response of a unidirectional MMC with an assumed periodic hexagonal packing of fibers in the commercial software package ABAQUS/Explicit. The arrangement of the fibers is expected to affect the elastic-plastic response of the composite, particularly when comparing rectangular and hexagonal packing e.g. [11]. The microstructure of the material in Figure 1 is best represented as a hexagonal or random arrangement of fibers for which prior modeling efforts have shown that elastic-plastic predictions generated from hexagonal and random arrangements to be similar [42-43]. The unit cell has two constituents: Al_2O_3 fibers and Al-2%Cu alloy matrix. The length (L_{uc}) and width (w_{uc}) of the cell are 26.64 and 15.38 μm respectively. Since the fibers are very long with respect to their diameter, the out of plane depth of the cell (t_{uc}) was not very critical as it was assumed to be 1.0 μm with periodic boundary conditions applied. The fibers were assumed to have a circular cross section with a diameter of 11.98 μm giving the cell a volume fraction of 0.55. The response of the composite was modeled under uniaxial tension loading conditions in the axial (along fiber axis) and transverse (loading perpendicular to fiber axis) directions (Figure 4). The fibers are considered isotropic and modeled as elastic to failure with a density of 3.9 g/cm^3 , Young's modulus of 390 GPa [44] and an assumed Poisson's ratio of 0.23. The average filament strength of 2.97 GPa obtained experimentally [39] was used as the failure criteria to initiate damage in the fibers. The Al-2%Cu matrix was modeled as isotropic elastic-plastic with an equivalent plastic strain failure criterion. A Young's modulus of 70.9 GPa, Poisson's ratio of 0.3, and density of 2.8 g/cm^3 was used for the elastic properties. From the nano-hardness experiments it was found that the hardness of the matrix in-situ within the composite was 1.28 times higher than that of the unreinforced matrix. Based on this finding, the plastic constitutive

behavior used for the numerical modeling was obtained by linearly scaling the average uniaxial tensile stress-strain curve of the unreinforced matrix by a factor of 1.28 (Figure 3). The strain to failure was taken as the average value obtained from uniaxial tension tests and a value of 0.127 was used. Periodic boundary conditions were applied to all surfaces of the unidirectional composite unit cell. Velocity boundary conditions were applied to the appropriate boundary to deform the cell in uniaxial tension in the transverse (Figure 4A) and axial (Figure 4B) directions. The effective engineering stress ($\bar{\sigma}_e$) - engineering strain ($\bar{\varepsilon}_e$) curve of the composite was then calculated from the displacement (u_i) and reaction force (F_i) of the boundary. Using the transverse case as an example:

$$\bar{\varepsilon}_e = \frac{u_i}{w_{uc}} \quad (1)$$

$$\bar{\sigma}_e = \frac{\sum F_i}{w_{uc} t_{uc}} \quad (2)$$

Cohesive zones were used to model the behavior and failure of the interface between the Al_2O_3 fibers and the Al-2%Cu matrix. The constitutive behavior of the cohesive elements was modeled using a traction-separation law, which is shown schematically in Figure 5. This model assumes linear elastic behavior up to the point at which the interface strength is reached and at which damage is initiated. Initiation of damage, a scalar damage evolution variable (D) is introduced which describes the rate at which the material stiffness is degraded. In this work a maximum nominal stress criteria was considered for damage initiation. This criterion is expressed as:

$$\max \left\{ \frac{\langle t_n \rangle}{t_n^o}, \frac{t_s}{t_s^o}, \frac{t_t}{t_t^o} \right\} \quad (3)$$

where t_n , t_s , and t_t are the pure normal, and two shear components of stress at the interface, and t_n^o , t_s^o , t_t^o , are the maximum allowable values of the nominal stress. The Macaulay brackets $\langle \rangle$ indicate that a purely compressive stress state does not initiate damage.

The damage evolution variable of the cohesive zone elements evolves monotonically from 0 to 1. In this work a linear degradation in stiffness is assumed in which the normal and shear components of the traction-separation material behavior are evolved by the following relations:

$$t_n = \begin{cases} (1-D)\bar{t}_n, & \bar{t}_n \geq 0 \\ \bar{t}_n, & \text{otherwise} \end{cases}, \quad (4)$$

$$t_s = (1-D)\bar{t}_s,$$

$$t_t = (1-D)\bar{t}_t$$

where $\bar{t}_n$, $\bar{t}_s$, and $\bar{t}_t$ are the stress components of the undamaged material and damage is assumed to only progress when normal tractions are tensile ($\bar{t}_n \geq 0$). Damage evolution is defined in the model based on the energy dissipated as a result of the damage process and is equal to the area under the traction-separation curve (G_c). These parameters can be extremely difficult to measure directly from experiments and in this work the cohesive zone parameters, t_n^o and G_c were inversely determined by calibrating the simulation results to fit the experimentally obtained stress-strain curve for the transverse uniaxial tension test in which failure is dominated

almost exclusively by fiber-matrix interface separation which is observed in the micrograph of the post-test fracture surface shown in Figure 6A. When the cohesive elements are completely damaged they are removed from the mesh and frictionless contact is assumed between the fibers and the matrix.

3.2 Unidirectional MMC results

3.2.1 Experimental observations and deformation behavior

As expected, the experimental results are transversely isotropic between the transverse (Figure 7) and axial (Figure 8) loading directions. The transverse direction exhibits significant plasticity while in the axial fiber direction the MMC exhibits a bi-linear behavior up to ultimate failure. In both testing orientations there is considerable scatter in the strength and strain to failure properties which reflects the brittle nature of these MMCs and their sensitivity to local flaws. The elastic modulus (calculated at 0.05% strain), strength, and strain to failure of the unidirectional MMC are summarized in Table 1. The average strength was lower by a factor of 6.3 in the transverse case versus the axial case which is not surprising as the high strength ceramic fibers bear most of the load in the axial case. However the factor of 1.75 lower strain to failure observed in the transverse case is somewhat surprising as some moderate ductility would be expected due to the load being carried by the matrix in this orientation which should be able to accommodate larger plastic strains than the Nextel fibers in the axial orientation. This result suggests that the fiber-matrix interface is very weak and the fiber pull out, observed in the experimental fracture surface (Figure 6A), dominates the behavior in transverse loading.

The bi-linear behavior in the axial loading case is commonly observed in fiber reinforced MMCs [8] due to local plastic deformation of the aluminum matrix. After a very small initial

elastic period up to a strain of ~0.06-0.08%, the slope is reduced by 14%. This transition occurs when the matrix begins to plastically deform. A similar transition strain of 0.06% has been previously reported for a unidirectional Nextel 610TM reinforced aluminum alloy [27]. As global strain continues to increase the plastic strain in the matrix monotonically increases due to the strain hardening behavior of the Al-2%Cu at a rate lower than that at which the stress increases in the fibers which continue to bear the majority of the load. The linear behavior is then maintained until ultimate failure which occurs by brittle fracture when the strength of the fibers is reached and before the plastic strain saturates in the matrix. The variability in the strength and strain to failure is expected based on previously published data on unidirectional reinforced MMC [25, 45] due to the brittle nature of the Al₂O₃ fibers which follow a Weibull type distribution of strength due to their inherent sensitivity to internal flaws [39-41]. As such, the mean strength of the composite will be limited by the lower strength fibers which will fracture first [18, 28].

3.2.2 Unidirectional MMC simulation results and discussion

The cohesive zone parameters for the interfacial strength were determined by iteratively simulating the transverse uniaxial tension test to fit the result to the average experimental curve (Figure 7). Based on this process the final stress-strain curve obtained using cohesive parameters of: $t_n^o=155$ MPa, $t_s^o=t_t^o=89.5$ MPa, and $G_c=75$ J/m² is presented in Figures 7 and 9 along with the average experimental result (Figure 9) for comparison. To quantify the impact of our assumption of using scaled matrix strength based on the in-situ hardness measurements, a simulation result which uses the as-cast matrix properties is also presented for comparison in Figure 7. In this case of transverse tension we see that the hardening of the matrix in-situ due to

thermal mismatch stresses does not significantly impact the homogenized composite behavior which is dominated by interfacial debonding.

The result from an elastoplastic simulation without cohesive interface failure (perfect bonding assumed) is also included for reference in Figure 9. Over the range of strains presented, the elastoplastic simulation without interface failure appears nearly elastic as the fibers are able to bear most of the load when the interface remains intact. This result represents an upper bound on the strength and demonstrates the significant reduction in overall strength contributed by the interfacial failure mechanism. Please note that the simulation results for the perfectly bonded case are truncated to the range of strains in which the experimental specimens fractured and do not represent failure in the model which would occur at a strain greater than 1% which is clearly unrealistic based on experimental observations. We are able to quantify the amount of strain energy dissipated by interfacial damage by comparing the areas under the predicted stress-strain curves for the transverse simulations with and without interfacial damage considered. We find that interface separation accounts for 53% of the total strain energy dissipated by the material. The simulation with cohesive interfaces agrees well with the experiment in the elastic regions and in the later stage of deformation, including failure which occurs at an applied strain of 0.36%. The strength and failure strain were well within the range of experimentally measured values. There is however, a deviation in the initial yielding portion of the stress-strain curve with the simulation over predicting the experiments slightly. A limitation of the model is that damage in the periodic cell is described by the debonding of few fibers whereas in the experiment damage accumulates over a larger volume of fibers which would likely lead to a more gradual evolution of damage. The deviation can also at least be partially explained by processing induced defects in the material such as micro porosity and intermetallic inclusions (Figure 1) that

are not accounted for in the present model. These defects would be expected to reduce the effective yield stress of the MMC significantly [28, 46]. Defects at the fiber-matrix interfaces are particularly problematic as these locations will generally be the location of stress concentrations and the areas most likely to initiate interfacial damage and plastic deformation of the matrix. As such, we emphasize that the cohesive parameters calculated in this work represent the “effective” interface which includes these casting related processing defects of porosity and intermetallics.

The cohesive parameters obtained from calibration of the transverse tension test were used to simulate the response of the material in the fiber (axial) orientation. These simulation results are presented in Figure 8 along with the experimental results. Similar to the transverse loading case, the elastoplastic simulation considering a perfectly bonded interface between the fibers and matrix (no cohesive interface failure) is included for reference as an upper bound (Figure 10). The transverse and axial results are presented together on the same scale in Figure 11 to clearly illustrate the transversely isotropic properties in the two loading directions. Compared to the perfectly bonded simulation, the axial response is softened by interface failure and progressive damage accumulation which results in a simulation result that is more closely aligned with the experimental results which show clear evidence of interfacial debonding (Figure 9). Again we compare the strain energy under the stress-strain curves to quantify the contribution of interfacial damage to the overall response and find that shear damage of the interfaces accounts for 8% of the total strain energy dissipated in axial loading. In accordance with the experiment, a bi-linear response is predicted with a reduction in slope of 15% at a strain of 0.3%. Although the slope change occurs at a slightly smaller strain in the simulation than in the experiment, the magnitude of the reduction is accurately predicted. The FEM simulations

show that the matrix first yields in axial loading at very small strain of 0.12% which is below the point where the slope transition is observed. Failure in the axial direction is dominated by brittle fiber fracture (Figure 6B); however interface damage due to interfacial shear stresses is also observed as an active mechanism (Figure 6B) and contributes to the overall composite response. In the axial loading condition when the interface fails, the fibers still carry most of the load and the drop off in performance is not nearly as dramatic as in the transverse loading case. Ductile matrix failure is not a significant failure mechanism in this system. The MMC fails at a global applied strain of 0.64% while the maximum local plastic strain in the aluminum matrix is less than 1% which is well below the 12.7% critical strain specified for the Al-2%Cu. These results demonstrate that the parameters calibrated from the transverse tension experiment capture the relevant deformation mechanisms of the MMC and provide a reasonably accurate prediction of the average composite response when the dominant failure mode in transverse loading is through progressive damage and separation of fiber-matrix interfaces. These results provide a first order validation of the calculated cohesive parameters but additional experimentation should be considered in the future to more completely characterize and quantify the constitutive response of the metal-fiber interface particularly with regards to delineating mode I, II, and III fracture behavior. Additionally, it is important to note that the current simulations use a fixed average fiber strength which will only predict a deterministic single failure strain of the MMC. In this instance, the predicted failure strain was at the high end of the experimentally observed range. In order to predict statistically representative MMC properties it is necessary to include variation of the fiber properties in the constitutive model. An additional axial loading case was simulated to explore the effect of fiber strength on the composite properties and quantify the upper bound on properties that could be obtained by this material system by considering only high strength

fibers. The simulation was conducted assuming a fiber strength of 4 GPa which corresponds to fiber strengths in the upper range of experimentally measured values [39-41]. The simulation results (Figure 10) match exactly the previous simulation results but extend the curve and show that a 60% increase in MMC strength and a 62.5% increase in strain to failure of the MMC could be obtained by increasing the average fiber strength by 35%.

4. Numerical modeling of 2D weave fabric reinforced MMC

4.1 Model setup

A 3D unit cell consisting of MMC yarns and aluminum matrix was used to model the deformation response and predict homogenized macroscopic composite properties for the 2D fabric MMC which is generalized as a plain weave architecture. A schematic of the cell and yarn geometries are shown in Figure 12. The length (L_c), width (w_c), and height (h_c) of the cell were 2.96, 2.96, and 0.316 mm respectively. The dimensions of the cell were determined from the microstructure in Figure 2A where L_c and w_c are based on the distance it takes the warp yarns to repeat an undulation which is based on their crimp angle and is relatively small in this case. The tows were approximated as having cross sections with width (w_t) and height (h_t) of 1.08 and 0.136 mm respectively with rounded corners of radius 0.063 mm. The dimensions of the tows were obtained from averaging the dimensions of the yarn cross sections in Figure 2A. Based on this geometry the volume fraction of yarns in the cell is 0.55. The yarns are unidirectional MMC with a volume fraction of fiber of 0.55 which yields an equivalent volume fraction of fiber in the fabric unit cell of 0.30 to correspond with the experimental material. The unidirectional MMC yarns have transversely isotropic material behavior which requires defining local coordinate systems for each yarn. Furthermore, the crimp of the tows causes the local fiber axis to rotate with respect to the global coordinate system along the length of the yarn which causes the

effective modulus in the global direction to vary [47]. Therefore it was necessary to assign a local coordinate system to each element such that the local fiber axis is always parallel to the local 1 direction along the length of the yarn (Figure 13). The properties are transversely isotropic in the plane of the local 2 and 3 directions and correspond to the transverse properties of the unidirectional MMC, while the constitutive behavior of the local 1 direction corresponds to the axial orientation. The overall transversely isotropic elastic-plastic constitutive behavior of the MMC tows in the fabric unit cell was taken from the predicted simulation results for the unidirectional MMC (Figure 11) to validate the effectiveness of this hierarchical multi-scale modeling approach. Elasticity of the unidirectional composite was defined as transversely isotropic with modulus of the axial (local 1) direction of 246.5 GPa and 148.3 GPa in the transverse (local 2 and 3) directions. The shear moduli were all assumed to be 46.3 GPa based on tests conducted on unidirectional specimens tested at 45° to the fiber axis [39]. In order to implement plasticity in the transverse direction it was necessary to use the Hill potential for anisotropic plasticity in ABAQUS [48]. The plastic portion of the unidirectional transverse stress-strain curve was input directly in tabular form with an initial yield strength of 87.5 MPa to capture the strain hardening behavior and used as the reference stress. The stress ratio for the Hill potential to define plasticity in the axial direction (σ_{11}) was calculated from the ratio of the axial and transverse curves in Figure 11, while the ratios for the other stress components (σ_{22} , σ_{33} , σ_{12} , σ_{13} , and σ_{23}) were all set equal to one. An equivalent plastic strain failure criterion was used to ensure that yarn failure in axial loading occurred at a stress of 1487 MPa and at 182.3 MPa in the transverse directions. Implementation of the unidirectional MMC properties in ABAQUS/Explicit was conducted using a single element model subjected to uniaxial loading in each of the 3 principal directions and verified to successfully reproduce the curves in Figure 11.

Each MMC tow consists of 4012 solid hexagonal stress elements and the aluminum matrix was composed of 90,612 solid tetragonal stress elements, giving a total of 106,660 solid stress elements (Figure 12C). Homogeneous boundary conditions were applied to opposite sides of the cell as indicated in Figure 12C, with one side being restrained from motion in the loading direction and a uniform displacement boundary condition to deform the cell in uniaxial tension in the plane of weaving. All other surfaces were free to deform. Simulations were conducted with and without cohesive zone layers to model the interface between the tows and the matrix and quantify the contribution of interfacial decohesion to overall deformation response of the fabric MMC. The Al-2%Cu matrix was modeled the same as in the unidirectional modal discussed in section 3.1 above, as isotropic elastic-plastic with an equivalent plastic strain failure criterion (equivalent plastic strain at failure = 0.127). The same formulation outlined above and the cohesive parameters calibrated from the unidirectional MMC simulations were used. When the cohesive elements are completely damaged they are removed from the mesh and frictionless contact is assumed between the tows and the matrix.

4.2 Fabric MMC results and discussion

A total of 8 fabric reinforced MMC tensile specimens were tested to failure and found to have an average strength of 326.4 ± 27.6 MPa and average elongation of $0.41 \pm 0.04\%$. There was significant scatter in the experimental data due to the brittle nature of the MMC with strengths ranging from 280-353 MPa and failure strains ranging from 0.34-0.46%. The in-plane tensile behavior of the 2D weave fabric MMC is bi-linear to failure (Figure 14). After an initial elastic period up to a strain of ~ 0.04 - 0.05% there is a reduction in slope of approximately 50%. The reduction in slope is 3.3 times greater than in the unidirectional material which is due to the larger volume of matrix in the fabric MMC. The average in-plane elastic modulus measured

experimentally in the initial portion of loading at a strain of 0.05% was 139.2 GPa. SEM micrographs of the post tensile test fracture surfaces are presented in Figure 15. From these micrographs multiple deformation and damage mechanisms can be observed at different length scales. The dominant failure mechanisms were axial fiber fracture and transverse fiber decohesion. In the lowest magnification micrograph (Figure 15A) the fabric architecture is seen along with failure of warp and weft tows. The fracture surface has a large degree of topology where tows have fractured heterogeneously suggesting that the composite was able to distribute load to intact tows as damage progressed prior to ultimate failure. In Figure 15B, separation of tows from the matrix was observed indicating that failure of the interface between the tows and matrix acts as an active damage mechanism at the fabric scale. Interfacial failure on the tow level can also be observed in a polished cross section taken away from the fracture surface (Figure 16). In the two higher magnification micrographs (Figure 15 C, D) sub-yarn scale damage mechanisms are more clearly observed. Fracture of individual fibers in the axial (D) and transverse (C) orientations and interfacial decohesion in both the axial (D) and transverse (C, D) orientations is observed. Little evidence of ductile matrix failure or plasticity is observed in this composite system signifying that the strength and elongation is dictated entirely by the inherent strength of the fibers and the quality of the bond between the fibers and matrix.

Three separate simulations were conducted to explore and quantify the effects of individual MMC deformation and damage mechanisms on the overall composite response. The predicted stress-strain curves up to, and including material failure, are presented along with the experimental results in Figure 14. The first and simplest simulation considered the MMC yarn as transversely isotropic and elastic to failure with the transverse plasticity not considered. By comparing the results of this simulation with a simulation including transverse plasticity

determined from studying the unidirectional MMC we are able to quantify the contribution of transverse plasticity to the overall response. Finally, we add cohesive interfaces between the tows and matrix which allows quantification of the amount of strain energy that is dissipated through interface decohesion and damage. All three simulations predict the bi-linear composite behavior with the change in slope occurring at a strain of approximately 0.07% which is at a slightly higher strain than observed experimentally. The simulation results show good agreement with the modulus at small strains with a predicted value of 142 GPa at a strain of 0.05%, which is within 2% of the experimental value. While the simulation without transverse plasticity accurately predicts the modulus at small strain, the stress is over predicted with deviation that increases compared to the experimental result as the strain increases and plasticity and damage mechanisms become active processes. When plasticity of the transverse tows is not considered, the stresses and plastic strains increase at a higher rate in the axial tows (Figure 17) and they are forced to accommodate the majority of the strain energy input into the composite. This results in the ultimate strength of the axial tows being reached at a relatively small strain (19% lower than the experimental average) and results in brittle fracture of the axial tows with little transverse damage predicted. This failure mode also does not correspond to that observed experimentally (Figure 15) in which significant failure of transverse tows is shown. These results clearly demonstrate the importance of transverse tow plasticity in the tensile response of fabric metal matrix composites.

Inclusion of transverse plasticity of the tows in the simulation softens the predicted response and shows closer agreement to the experiment, especially at larger strains. For example, at a strain of 0.3%, the stresses are 242, 263, and 318 MPa, in the experiment, simulation with transverse plasticity, and the simulation without transverse plasticity

respectively. Additionally the strain to failure is increased by 17% compared to the simulation without transverse plasticity considered and falls within the range of experimental failure strains and is 5% below the experimental average. If we compare the strain energy to failure (area under the curve) we see that including transverse plasticity results in 13.5% more total energy absorption of the composite. These results clearly demonstrate the significance of the role played by transverse plasticity of the tows perpendicular to the loading axis in the macroscopic deformation response of the MMC. The transverse tows contribute significantly because of the relatively effective interface that forms between the tows and the matrix during infiltration casting which is able to transfer a substantial portion of the load that is primarily handled by the axial tows. The simulation which considers transverse tow plasticity shows reasonable agreement with the experiment in the initial elastic portion and at larger strains, however there is a significant portion of the inelastic stress-strain response that is not accurately predicted between strains of ~ 0.05 - 0.3% in which the simulation stress is significantly higher than the experiment suggesting additional damage processes are occurring in the material. Confirmation that interfacial debonding is occurring is observed in the cross section of the post-fracture test specimen in a region away from the fracture surface (Figure 16). In order to address this deficiency in the modeling, cohesive zones were included in the simulation to model the interface between the MMC tows and the matrix. As shown in Figure 14, the simulations including cohesive interfaces accurately predict the macroscopic constitutive response of the fabric MMC over the entire stress-strain curve including elasticity, plasticity, and ultimate failure. A summary of the experimental and simulation results is presented in Table 2. By calculating the area under each predicted stress-strain curve up to failure we can compare the strain energy dissipated in the three simulations and quantify the contribution of each

deformation mechanism to the overall composite response. First, we compare the simulations without cohesive interfaces and see that the simulation which includes transverse plasticity of the tows dissipates 13% more strain energy to failure than the model which only considers the properties of the axial tows which bear the majority of the load. This corresponds to 12% of the total in plane strain energy capacity of the material. The cohesive interfaces on the fabric scale dissipate an additional 10% of strain energy which is 9% of the total in plane energy dissipated by the composite to failure.

Stress concentrations at the corners of the tows have two primary effects on the local deformation characteristics. The first is the initiation and increase in local plastic deformation of the matrix. The second is that the interface will first separate at these locations at relatively small strains due to the high local stresses which introduce damage in the MMC when cohesive zones are present. As the interfaces progressively debond the load is redistributed and the axial tows (Figure 17A), and the matrix, are required to carry more of the load than in the case where interfacial failure is not considered. This results in increased plastic strain in the matrix (Figure 17C). As the matrix is forced to accommodate more plastic strain when the interface becomes damaged, less load, and subsequently strain, is imparted to the relatively weak transverse tows. This has the net effect of extending the overall elongation of the MMC to 0.437% which is 11% greater than the simulation without cohesive zones. Similar to the simulation without interface damage, this result is within the range of experimentally observed values and 5% greater than the experimental average.

The observation that the simulation without cohesive interfaces between the tows and matrix matches well with the experiments in the late stage of deformation suggests that ultimate failure is dictated mostly by progressive damage and failure of the transverse tows and that

decohesion of the tows from the matrix contributes mainly in the early to mid stage of the process. When cohesive zones are considered little damage to the transverse tows is observed in the simulation. This is due to the loss of load transfer and subsequent reduction in plastic deformation in the transverse tows when the cohesive interface fails in tension. This result suggests that the interface on the tow scale has a higher effective strength than the interface between individual fibers within a tow for which the present cohesive parameters were determined. One possible explanation for this discrepancy is that local stresses within the yarn are higher due to the higher local volume fraction and packing of fibers within the tow which would be expected to decrease the effective interfacial shear strength between the fiber and matrix. This effect would be exacerbated by local porosity at the interface which cause the interfacial cohesive parameters calibrated from the unidirectional transverse tension to underestimate the effective cohesive parameters to be used at the fabric scale. New cohesive parameters can be calibrated to give better qualitative agreement between simulation and experimental fracture surfaces however the fact that the stress-strain response is accurately predicted is a significant result in itself as it demonstrates that the underlying damage mechanisms are quantitatively accounted for in terms of total strain energy dissipated through interfacial separation at the two length scales considered in the present numerical framework.

5. Conclusions

A multi-scale finite element micromechanical modeling framework was developed to study the complete transversely isotropic tensile deformation response, including transverse plasticity, damage, and progressive failure, of continuous ceramic fiber and woven fabric reinforced metal matrix composites. Deformation mechanisms were experimentally characterized in unidirectional and woven composites and incorporated into the numerical

simulations to yield accurate predictions of the transversely isotropic tensile properties of unidirectional and woven fabric MMC starting from only the constitutive behavior of the individual ceramic fibers and aluminum matrix. The model incorporates the three dominant failure mechanisms present in MMC systems: ductile matrix failure, cohesive interfacial separation, and brittle fiber fracture, of which the contributions from each were quantified. It was demonstrated that interfacial separation of fibers from the matrix is the dominant damage mechanism in transverse loading of the unidirectional MMC and the strength and ductility of the matrix plays very little role in the homogenized constitutive response. Validated cohesive parameters for the fiber-matrix interface were used to quantify the strain energy dissipated by decohesion in uniaxial loading of the unidirectional MMC as well as being suitable for describing the strength of the interface between tows and matrix in the fabric MMC model. The contribution of each deformation mechanism to the overall strain energy capacity of the 2D weave composite was quantified and it was shown that while axial tows bear the majority of the load, transverse plasticity and interfacial decohesion of tows account for 21% of the total strain energy dissipated during in plane tensile deformation.

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Figure 1. SEM micrograph of unidirectional Nextel 610[™]-Al-2%Cu MMC. Note porosity (black) and intermetallic θ Al₂Cu precipitates (white) at fiber-matrix interfaces.

Figure 2. SEM micrographs of 2D weave Nextel fabric (cross section normal to plane of weave) reinforced MMC at low magnification (left) showing yarns and weave architecture, and high (right) magnification showing sub yarn fiber structure.

Figure 3. Experimental uniaxial tension plastic stress-strain curve and constitutive plastic stress-strain curve used in numerical modeling for Al-2%Cu obtained by scaling experimentally obtained curve by in-situ hardness measurements.

Figure 4. 3D periodic cell model used in present simulations for transverse (A) and axial fiber direction (B) loadings. Finite element mesh is shown in (B).

Figure 5. Schematic of traction-separation law used to model progressive damage and failure of matrix-fiber interfaces.

Figure 6. Fracture surfaces of transverse (A) and axial (B) tension test specimens of unidirectional continuous Al₂O₃ fiber reinforced Al-2%Cu.

Figure 7. Experimental stress-strain curves for transverse tension test for all specimens showing significant scatter in strength and strain to failure along with simulation predictions for models with calibrated interface strength for model using as-cast matrix properties (solid black line) and model using matrix properties scaled using in-situ hardness measurements in the composite (dashed black line).

Figure 8. Experimental tensile stress-strain curves for all specimens showing significant scatter in strength and strain to failure along with simulation prediction (dashed line) for model with calibrated interface strength for axial tension test.

Figure 9. Comparison of average experimental results with elastoplastic simulation predictions for perfectly bonded example (no interface failure) and elastoplastic simulation predictions with calibrated interface strength for transverse tension test of unidirectional MMC (note that perfectly bonded simulation is not to failure).

Figure 10. Comparison of average experimental results (dashed line represents range of failure strains) with simulation predictions for perfectly bonded (no interface failure) and simulation predictions with calibrated interface strength for axial tension test.

Figure 11. Comparison of experimental and numerical generated stress-strain curves for unidirectional continuous Al_2O_3 fiber reinforced Al-2%Cu in transverse and axial loading orientations.

Figure 12. Components of 2D plain weave fabric unit cell model: tow geometry (A), unit cell schematic with translucent matrix to visualize internal tows and weave architecture (B), finite element mesh with homogeneous boundary conditions (C), and tows with cohesive zone (green) interface layer (D).

Figure 13. MMC yarn geometry showing local coordinate systems for definition of transversely isotropic constitutive material behavior in woven fabric unit cell. The local 1 direction corresponds to the fiber axis of the unidirectional MMC and remains parallel to the crimp of the tow, while the 2 and 3 directions correspond to the transverse direction.

Figure 14. Comparison of representative experimental (solid line) results with simulation predictions for 2D weave fabric MMC. Simulation results for a model using only the axial properties of the unidirectional MMC, a model including full orthotropic elastic and transverse plasticity, and an orthotropic elastic-plastic model including cohesive interfacial separation are shown.

Figure 15. Post quasi-static tensile test SEM micrographs of 2D fabric MMC fracture surfaces at different magnifications showing multiple deformation and damage mechanisms including: fracture of axial tows, pull out of axial tows, transverse fiber pull out, axial fiber fracture, transverse fiber fracture, and pull out of axial fibers.

Figure 16. SEM micrograph of polished cross section of fabric reinforced MMC post-fracture tensile specimen. Micrograph was taken away from the fracture surface and shows evidence of individual filament (top) and tow level (lower) decohesion. Arrows indicate direction of loading.

Figure 17. 2D weave fabric simulation results: axial stress (MPa) in axial tows (A), equivalent plastic strain in transverse tows (B), and equivalent plastic strain in the aluminum matrix (3/4 cross section, C) for simulations without cohesive interfaces between tows and matrix (LEFT) and simulations with cohesive interfaces included (RIGHT). Arrows in (A) indicate direction of loading. Contours taken at strain corresponding to maximum load just prior to complete failure which is at a strain of 0.0039 and 0.0047 for the simulations without and with cohesive interfaces respectively.

Table 1. Summary of average experimental and numerically predicted mechanical properties of unidirectional Al₂O₃ fiber MMC

Nextel unidirectional MMC			
	Modulus (GPa)	Strength (MPa)	Elongation (%)
Axial - experiment	219.1 ± 24.8	1280 ± 176.2	0.56 ± 0.093
Axial - simulation	246.5	1487	0.64
Δ (%)	11.1	16.2	14.3
Transverse - experiment	144.1 ± 12.9	189.9 ± 11.2	0.324 ± 0.082
Transverse - simulation	148.3	182	0.36
Δ (%)	2.9	4.0	11.1

Table 2. Summary of in-plane properties of woven Al₂O₃ fabric MMC obtained from experiments and simulations.

Nextel Fabric MMC			
	Modulus (GPa)	Strength (MPa)	Elongation (%)
Experiment (N=8)	139.2 ± 8.72	326.4 ± 27.6	0.413 ± 0.043
Simulation with cohesive interfaces	142	327	0.44
Δ (%)	1.9	0.1	5.7