FIBER FRACTURE DURING THE CONSOLIDATION OF METAL MATRIX COMPOSITES

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Abstract—A detailed study of conditions leading to fiber fracture during the consolidation of Ti-14wt%Al-21wt%Nb/SiC (SCS-6) composite monotapes has been conducted. For this continuous fiber reinforced composite system, the incidence of fracture increases with consolidation rate at higher process temperatures. Increasing consolidation temperature at a fixed pressure reduces the number of breaks per unit length of fiber. Examination of partially densified compacts has revealed the existence of significant fiber bending and ultimately fracture due to monotape surface roughness (asperities) which places the fibers in three point bending. A representative volume element has been defined for the consolidating lay-up and its response analyzed to predict the fiber deflection (and hence probability of failure) when the surface asperities deform either by plasticity or by steady state creep. The relationships between fiber fracture and process conditions predicted using the volume element are similar to those observed experimentally. The cell analysis suggests that fiber fracture is decreased by increases in fiber stiffness, strength, and diameter and by decreases in matrix yield and creep strength and monotape surface roughness.

Résumé—Une étude détaillée des conditions qui mènent à la fissure des fibres pendant la consolidation des monotapes composés Ti–14wt%Al–21wt%Nb/SiC (SCS-6) a été effectuée. Pour ce système composé réinforcé de fibres continues, l’incidence de fissure augmente avec le taux de consolidation à des températures de processus élevées. L’augmentation de la température de consolidation à une pression fixée réduit le nombre de fissures par unité de longueur dans les fibres. Une examination des compacts partiellement densifiés a révélé l’existence d’une courbature importante des fibres et une éventuelle fissure due aux âprets de la surface du monotape, ce qui fait courber les fibres à un point lorsque les bouts des fibres restent fixes. Un élément de volume représentatif a été défini pour le composé consolidé et sa réponse a été analysée pour prédire la déflexion des fibres (et donc la probabilité d’échec) lors de la déformation des âprets de la surface soit par la plasticité soit par la déformation non-temporelle à haute température (steady state creep). Les relations prédites entre les fissures dans les fibres et les conditions des processus en utilisant l’élément de volume sont semblables à celles qui ont été observées dans les expériences. L’analyse des cellules suggère que la fissure des fibres décroit avec des augmentations en rigidité, fortitude et diamètre, et avec des décroits en résistance totale de la matrice à la déformation et en résistance de celle-ci à la déformation à haute température ainsi qu’en âpreté de la surface du monotape.


1. INTRODUCTION

The structural material needs of the aerospace industry have stimulated the development of high-temperature materials for many years. Today, materials research for these and other high-temperature applications is focusing upon continuous fiber (e.g. silicon carbide and aluminum oxide) reinforced metal (or intermetallic) matrix composites because of their attractive combination of specific stiffness and
strength both at ambient and high temperatures [1, 2]. Numerous processing approaches are being explored for combining fibers and matrices into desirable end products [3–5]. One increasingly favored approach utilizes thin (200–550 μm) plasma-sprayed monotape sheets consisting of uniformly spaced, continuous, parallel fibers in an alloy matrix [6]. Lay-ups of these monotapes can be consolidated to form near net shape composite components by either hot isostatic or vacuum hot pressing (HIP/VHP) [6]. The monotapes are produced by depositing plasma-melted alloy droplets onto a unidirectional array of ceramic fibers attached to a substrate. The tapes thus formed (Fig. 1) typically have one smooth surface (the one in contact with the substrate), one rough surface (the last to solidify during deposition), and varying degrees of internal porosity.

Several studies have reported the existence of fiber fractures in numerous samples consolidated in this way [5, 7–10]. However, little work has been done to correlate fiber fracture with processing conditions, to identify specific fiber fracture mechanisms, or to develop predictive models of the damage process. The one partial exception is the related work of Tszeng et al. [11] on spherical powder–fiber compacts. They observed that, during consolidation, individual powder particles impinged upon fibers and caused them to bend and fracture. Based on these findings, Tszeng et al. modeled this fiber fracture process using a representative volume element consisting of an elastic fixed-fixed beam (the fiber) undergoing bending as a result of contact with a plastically deforming sphere (the powder particle). The force exerted by the deforming particle was found to be responsible for the fiber’s fracture. Although insightful, Tszeng’s study of the interaction between powder particles and individual fibers does not incorporate all of the phenomena that occur in the monotape lay-ups of interest here. For instance, it did not consider the role of time dependent (creep) deformations which are responsible for much of the densification during monotape consolidation [12]. Further, the study did not experimentally quantify the number of fiber fractures associated with various processing conditions and did not use the calculated stress in a single bent fiber to predict the number of fractures per length of fiber for a set of processing conditions (i.e. temperature and pressure).

In addition to limited discussion of the relationship between consolidation induced fiber fracture and processing conditions, the literature reports little about the effect of consolidation damage upon mechanical properties [13]. One might expect a significant degradation of the in-service properties of metal matrix composite (MMC) components. The possibility can be assessed for ambient temperature strength by a simple modification of Curtin’s recent analysis of the strength of a fiber reinforced composite [14, 15]. In the Curtin model, a Weibull strength distribution statistically characterizes a fiber’s strength. This can be easily modified to account for regions of the fibers which have zero strength because of processing fractures.

$$\phi (L, \sigma) = \exp \left[ -\frac{L}{L_0} \left( \frac{\sigma}{\sigma_0} \right)^m - L_0 \right].$$

Equation (1) is a modified Weibull distribution that characterizes the probability of fiber survival, $\phi (L, \sigma)$, after tensile testing of a sample to an applied stress $\sigma$, where $L$ is the total length of fiber being stressed, $L_0$ is the specimen gauge length, $\sigma_0$ is the reference stress for the fiber (the stress at which the survival probability is $1/e = 0.37$), $m$ is the experimentally determined Weibull modulus, and $\sigma$ is the number of fiber cracks per unit length of fiber after processing (but before subsequent testing). Using (1) for the fiber strength distribution, Curtin’s analysis leads to an expression for the composite’s ultimate tensile strength

$$\sigma_{\text{ult}} = f T \left[ 1 - \frac{r \sigma_0}{2 \pi L_0} \left( \frac{T}{\sigma_0} \right)^{m+1} - \frac{r \sigma_0^2}{2 \pi} \left( \frac{T}{\sigma_0} \right) \right]^{m+1}.$$

Figure 2 shows the predicted relative strength degradation that results from processing-induced fiber fractures in a typical fiber reinforced Ti–14wt%Al–21wt%Nb/SiC (SCS-6) composite.
When no fibers are broken ($\alpha = 0$), equation (2) yields a $\sigma_{\text{eff}} = 1.6$ GPa. The figure shows that fiber fracture densities of around 100/(m of fiber) result in a 20% loss of expected strength. One also anticipates that creep resistance would be even more significantly affected [16, 17]. To produce high quality composites by consolidation processing, it is important to elucidate the influence of processing conditions upon, and the mechanism responsible for, fiber damage. Then strategies for eliminating such damage can more confidently be pursued and more realistic models of the damage evolution developed.

The work reported here seeks to do this. It investigates the relationships between consolidation conditions and fiber fracture, it identifies the dominant mechanism of fiber fracture during monotape consolidation, and it proposes a unit cell micromechanical model for the fracture process that incorporates the competing effects of fiber and matrix deformation. This model is used to predict the relationships between process conditions and fiber fracture and to suggest process modifications which will reduce damage. A subsequent paper will develop a macroscopic (stochastic) predictive model that incorporates surface roughness and uses it to investigate the consolidation process more fully [18].

2. MATERIALS

Plasma-sprayed monotapes composed of a Ti–14wt%Al–21wt%Nb matrix and continuous SCS-6 silicon carbide fibers were provided by General Electric (GE) Aircraft Engines (Lynn, Mass.). They were produced using an inductively coupled plasma deposition (ICPD) process. In this process, plasma melted droplets of the metal matrix alloy are spray deposited onto a continuous fiber array prewound about a smooth cylindrical mandrel [19, 20]. Rotation and translation of the mandrel during spraying ensures uniform distribution of the matrix material. Two batches of spray-deposited monotape were used here; they are designated LPS-485 and LPS-669. For LPS-485 samples, the center-to-center fiber spacing was 230 $\mu$m and the foil thickness was 0.42 mm while the spacing for LPS-669 was 195 $\mu$m and the thickness 0.52 mm. Both batches of material had similar surface roughness. A detailed analysis of the tapes can be found in [21] and a complete characterization of the matrix microstructure and its evolution during processing in [22].

3. EXPERIMENTAL

The objectives of the experimental study were to (1) establish the conditions under which fiber fracture occurs during consolidation processing of plasma sprayed Ti–14Al–21Nb/SCS-6 composites, (2) identify the fiber fracture mechanism, and (3) correlate the number of fiber fractures with processing conditions [21]. Test specimens for consolidation by either hot isostatic or vacuum hot pressing were prepared from the monotapes by electrodischarge machining. They were ultrasonically cleaned, first in acetone, and then in methanol. Dissolution of the matrix showed that fewer than 10 fiber fractures/m resulted from sample preparation. For hot isostatic pressing, canisters were designed and constructed to accommodate a stack of five 3 cm x 3 cm monotape layers. The monotapes were stacked with their fibers parallel, sealed in canisters by electron beam welding in a 5 m torr vacuum, and then consolidated in an ABB Autoclave Systems QIH-15 HIP. Vacuum hot pressing of similar monotape lay-ups was conducted in an ATS Model 2710 stress relaxation machine. A fixture was designed and built to allow application of constrained uniaxial compression with stresses up to 80 MPa at temperatures as high as 1000°C in vacuum.

Hot isostatic press conditions were varied to investigate the influence of temperature and pressure on fiber fracture. For these experiments, LPS-485 samples were consolidated at temperatures of 825°C, 900°C, or 975°C and at pressures of 50, 100, or 200 MPa. Once attained, these conditions were maintained for half an hour. Samples were always heated to full temperature prior to the application of the full pressure. During heating it was necessary to maintain a minimum “prepressure” of 2 MPa at ambient temperature, increasing to 6.5 MPa at 900°C. For all HIP specimens, the heating rate was 20°C/min and the pressurization rate 85 MPa/h. Cooling and depressurization to ambient conditions was performed over a 2 h period.

Two sets of vacuum hot press tests were performed. The first set of experiments was conducted at low pressures (2, 3.5, and 6.9 MPa) and at room temperature and was intended to investigate the effect of HIP prepressurization upon fiber fracture. Both LPS-485 and 669 were investigated. The second set of experiments explored the influence of pressure ramp rate upon fiber fracture. For these experiments, LPS-669 samples were used. The temperature in the evacuated chamber was increased at 300°C/h until the specimens reached temperatures of 870°C, 955°C, or 1045°C. Compressive loads were then applied at crosshead displacement rates of 0.41, 0.80, or 1.67 cm/h, until the pressure on the sample reached 50 MPa. This pressure was held for 30 min before being rapidly removed. The specimen was then cooled at approximately 200–300°C/h.

Because the monotape foils were thin, the HIP specimens were found to deform significantly only in the through-thickness direction. In-plane compressive stresses merely constrained the specimen from lateral flow in the canister. Thus the stress state experienced by the hot isostatically pressed specimens approaches constrained uniaxial compression like that of vacuum hot pressing, and it is reasonable to presume that differences in the fiber fracture results between hot isostatic and vacuum hot press tests are
The results of these tests were used to generate a two parameter Weibull distribution of the fibers' strength.

4. RESULTS

Observations of fibers chemically removed from as-received monotapes revealed that no fibers were broken by the ICPD process. Their strengths were measured and fit to a Weibull distribution, Table 1. It can be seen that the fibers' strengths were not significantly affected by the ICPD process. Thus, any fiber damage observed after subsequent hot isostatic or vacuum hot pressing can confidently be attributed to either consolidation processing or sample preparation damage. The first two rows of Table 2 summarize the damage associated with sample preparation only. They show 5–7 preparation induced fiber breaks per meter of fiber and may be expected to contribute to the results of all tests. The remainder of Table 2 summarizes the effect of processing on fiber damage. It is apparent that every VHP and HIP consolidation cycle has resulted in fiber fracture. Sometimes, quite significant levels of fiber fracture accompany consolidation. Table 2 shows that the application of even small loads at room temperature leads to fiber breaks and that LPS-669 samples experienced significantly fewer fiber fractures than those prepared from LPS-485 at all stress levels. This is more clearly illustrated in Fig. 3 which plots the fracture density versus applied load during VHP for both materials.

<table>
<thead>
<tr>
<th>Test</th>
<th>Applied stress (MPa)</th>
<th>Temperature (°C)</th>
<th>Displacement rate (cm/h)</th>
<th>Soak time (h)</th>
<th>Relative density</th>
<th>Number of fibers in sample</th>
<th>Fracture density (m⁻¹)</th>
<th>Intact fibers (%)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Foil</td>
<td>0</td>
<td>20</td>
<td>NA</td>
<td>NA</td>
<td>NA</td>
<td>109</td>
<td>6.8</td>
<td>95.4</td>
</tr>
<tr>
<td>Foil</td>
<td>0</td>
<td>20</td>
<td>NA</td>
<td>NA</td>
<td>NA</td>
<td>136</td>
<td>5.0</td>
<td>95.6</td>
</tr>
</tbody>
</table>

Table 2. Fiber fracture results

Notes: *Batch LPS-485. †Batch LPS-669. ‡Results shown are average for 3–4 tests. †Fiber segments which are 95% of the sample length, or longer, are considered to be intact. *Sample subjected to a double soak. The sample was neither cooled nor depressurized between the two soaks. NA = not applicable. NM = not measured.
CONSOLIDATION OF METAL MATRIX COMPOSITES

Figure 4 shows the relationship between HIP consolidation temperature and pressure and the number of fiber fractures per meter for different pressures. It is evident that, as the processing temperature was increased (at a fixed pressure), the number of fiber fractures per meter decreased. For specimens processed at a fixed temperature, higher consolidation pressures increased the number of fiber fractures per meter—particularly at the lower processing temperatures. Figure 5 shows the effect of loading rate during VHP upon the number of fiber fractures per meter. At the two highest temperatures (955°C and 1045°C) the number of fractures per meter increased with rate of loading, but at the lower temperature (870°C) the number of breaks was generally independent of the applied displacement rate.

To investigate the cause of fiber fracture, a series of vacuum hot press experiments was interrupted shortly after the start of consolidation to prevent extensive metal flow from obscuring evidence of the failure mechanism. Examination of these partially consolidated specimens revealed bent and broken fibers. Figure 6(a) shows a bent and broken fiber from a fully consolidated HIP sample. The three fibers shown are all from the same monotape, and the middle fiber has clearly been subjected to localized bending. Figure 6(b) shows a side view of a typical fiber fracture in a sample interrupted 30 s after reaching a soak temperature and pressure of 900°C and 50 MPa. The fiber has fractured, apparently in bending at an asperity. Such fibers are observed to bend, curving about the asperity. In Fig. 6(b), the crack appears to have started at the top of the fiber (the tensilely loaded region) during bending and to have arrested in the bottom part of the fiber (the compressively loaded region).

5. DISCUSSION

The experimental results from this study provide the basis for a more quantitative, model-based investigation of fiber fracture during the hot isostatic or vacuum hot press consolidation of plasma sprayed monotapes. Once a model has been presented which reasonably describes the damage mechanism, it will
then be possible to identify material and processing parameters most strongly influencing fiber fracture, to reproduce experimentally observed fiber fracture trends, and to suggest improvements in the plasma deposition and consolidation processes. It is not our intent here to develop a model capable of accurate numerical prediction of cumulative fracture during consolidation. This requires the incorporation of stochastic features of spray-deposited surface topography and is dealt with in a separate paper [18]. Instead, we identify and analyze the behavior of a single representative volume element.

The magnitude of the deflection of a fiber segment spanning two adjacent asperities, and therefore the probability of fiber fracture, will be governed by a number of factors: the plastic and creep properties of the matrix (which vary with temperature, pressure, and density during consolidation), the stiffness, diameter, and strength of the fibers, and the size and spatial distribution of the surface asperities. The role that these parameters play in determining the rate of fiber fracture can be better understood by considering the micromechanical response for the representative volume element of Fig. 7. During consolidation, forces can be transmitted through the monotapes only at points where the rough surface asperities of one foil make contact with the surface of an adjacent foil. At the start of consolidation, contacts are widely spaced, resulting in nonuniform loading of the fibers and subsequent bending. Since the asperities are comparable in diameter to a fiber, this fracture mode may be expected to affect fibers individually [12]. For modeling purposes it is assumed that the asperities deform by a combination of plasticity and creep, and it is the asperities' resistance to this deformation that results in forces that cause fiber loading. The contact forces due to both mechanisms of deformation have been derived in earlier work on densification modeling [12]. The Appendix uses these results to calculate the stresses in the fiber due to bending and thus the probability of fiber failure (using the measured Weibull fiber strength statistics given in Table 1).

The model considers the asperity plastic flow (time-independent) and creep asperity deformation separately. It can first be used to examine low temperature pressurization, e.g. HIP prepressurization, where densification occurs only by plastic deformation. In this case the mid-point fiber deflection associated with plastic deformation (\(v_p\)) of a fiber segment is (see Appendix)

\[
  v_p = \frac{2(z_0 - z)}{(1 + 2\zeta)}
\]

where \(2z\) is the current volume element height (determined by the current macroscopic density), \(2z_0\) is the initial (undeformed) height, and \(\zeta\) is a proportionality constant equal to \(k_e/k_p\), where \(k_e\) is the fiber's elastic bend stiffness and \(k_p\) is a "plastic stiffness" proportional to the asperity (i.e. matrix) yield strength. As pressure is increased and densification progresses, the representative volume element in Fig. 7 is compacted. Thus \((z_0 - z)\) increases, and the fiber suffers an increasing deflection.

Appendix equations (A4) and (A5) show \(\zeta\) (and thus \(v_p\)) to be a function of the fibers' elastic modulus and diameter, the matrix yield strength, and the asperity spacing. If the asperity has a low yield strength or the fiber has a high elastic stiffness (i.e. \(k_p \ll k_e\)), the proportionality constant \(\zeta\) is large, the fiber's elastic deflection small, and the probability of fiber fracture low. The fiber's bend stiffness \(k_e\) can be raised by increasing its elastic modulus or diameter or by decreasing its length (i.e. decreasing the asperity spacing). This indicates that spray depositing finer
droplets (creating a more uniform surface) during monotape manufacture should result in a composite precursor which is more consolidation process-tolerant. Incorporation of the unit cell analysis into a macroscopic model accounting for the statistical surface features [18] permits numerical prediction of how much improvement results from a given refinement in the size of spray-deposited particles. Similarly, bending in the fiber can be reduced by decreasing the matrix yield strength (e.g. by increasing the processing temperature).

At higher temperatures ($T > 0.4 T_m$), asperities can also deform by time-dependent creep mechanisms [12]. For the case of power law creep (A11), the fiber’s midpoint deflection ($v_c$) is given by

$$v_c = \frac{2\varepsilon}{2\gamma} \left[ \exp(2\varepsilon \sqrt{2\gamma}) - 1 \right] \left[ \exp(2\varepsilon \sqrt{2\gamma}) + 1 \right]$$

where $\varepsilon$ is the cell compaction rate for a given applied pressure and temperature calculated using the densification model presented in [12], $t$ is the elapsed time since the start of densification, and $\gamma$ is a material constant proportional to $B$ in the Norton power law creep relation (A8).

Inspection of equation (4) and the expression for $B$ (see Appendix) reveals that as the temperature increases, $\gamma$ becomes large, causing $v_c$ to decrease ($v_c \to 0$ as $\gamma \to \infty$). Thus the probability of fracture decreases as the temperature increases. It is also apparent that decreasing $\varepsilon$ (i.e. the densification rate) decreases $v_c$ so that the probability of fracture is diminished with decreasing compaction rate. For this reason one expects less fiber damage if consolidation is allowed to proceed more slowly at first and at higher temperatures in order to accommodate densifying strains by creep deformation of the asperities.

The extent to which the model reflects the dominant physical mechanism of fiber fracture can be assessed from its ability to reproduce the experimentally observed trends. Combining the creep model results for fiber deflection (4) with the measured two parameter Weibull strength distribution allows prediction of the fiber fracture probability as a function of loading rate and consolidation temperature (A13).

Figure 8 shows the calculated probability of fracture (defined as the fraction of all possible fractures) using the creep parameters for Ti–14Al–21Nb given in Table 3. This figure shows that an increase in the displacement rate (densification rate) at a fixed temperature raises the probability of fiber fracture, in agreement with the experimentally observed trend at 955° and 1045°C (cf. Fig. 5). Another interpretation of Fig. 8 is that as the densification rate is increased, the consolidation temperature must be raised to avoid fracture. These conclusions are quite similar to those drawn by Tszeng et al. [11] on the basis of their work with the consolidation of metal powders around continuous fibers.

At lower processing temperature ($T \leq 0.4 T_m$) densification is accomplished primarily by rate independent plastic yielding of asperities. Using the plasticity model predicted fiber deflection (3) together with Weibull strength statistics for the fiber, leads to the result shown in Fig. 9. This figure illustrates that increasing the fiber/matrix stiffness ratio ($k_f/k_p$) leads to a decrease in the probability of fiber fracture for a fixed amount of densification (indicated by the unit cell true strain). The additional axes at the top of Fig. 9 are intended to indicate the influence of processing temperature on probability of fracture. They map the process temperature to the $k_f/k_p$ ratio through the known temperature dependence of the yield strength for the Ti–14Al–21Nb alloy [23–28]. Simply put, when the matrix has a low yield strength (i.e. when processing is conducted at elevated temperature) the asperities deform more readily, the fibers suffer less bending, and the probability of failure for any amount of densification decreases. This explains the experimental temperature trend

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**Table 3. Ti–14Al–21Nb material data [23–28]**

<table>
<thead>
<tr>
<th>Plasticity</th>
<th>Yield strength as $f(T)$ (MPa)</th>
</tr>
</thead>
<tbody>
<tr>
<td>● Room temperature</td>
<td>539.9</td>
</tr>
<tr>
<td>● Low temp range ($20 \leq T &lt; 800°C$)</td>
<td>$\sigma_f(T) = 539.9 - 0.32 \cdot T$</td>
</tr>
<tr>
<td>● High temp range ($T &gt; 800°C$)</td>
<td>$\sigma_f(T) = 1019.9 - 0.92 \cdot T$</td>
</tr>
</tbody>
</table>

| Creep | Power law creep constant, $B_0$ [1/(MPa$^{1.5}$ h)] | 10$^{-2}$ |
|       | Stress exponent, $n$ | 2.5 |
|       | Activation energy, $Q_a$ (kJ/mol) | 285 |

*The steady state creep rate is described by

$$\varepsilon_c = B_0 \cdot \exp \left( \frac{Q_a}{RT} \right) \sigma^*.$$
shown in Fig. 4. The increase in the number of fractures with pressure (for a fixed temperature) that is also shown in Fig. 4 is just a reflection of the larger unit cell strain (i.e. greater densification) (see Fig. 9). The cell analysis is also able to account for the disappearance of a fiber fracture-displacement (densification) rate dependence at low consolidation temperatures in Fig. 5. This occurs as the contribution to the deformation force by the plasticity mechanism (rate independent) rises above that due to creep (rate dependent).

One shortcoming of the creep and plasticity model appears to be its inability to account for the experimental trend of Fig. 3. In this figure a difference in the number of breaks for the two materials batches is shown. This can be rationalized by noting that the LPS-669 samples (a) had a smaller fiber spacing (which results in a more effective load transfer to adjacent fibers) and (b) a thicker matrix deposit (which distributes the load more uniformly along the fiber) [21]. The model does not consider either of these factors. Despite this shortcoming, it is still able to duplicate the main experimentally observed fracture trends.

This study has several implications for the manufacture and processing of metal matrix composites. First, precursors (e.g. monotapes and powders) need to have as high a starting density as possible with matrix characteristic dimensions (e.g. powder size, asperity size and spacing, and void diameter) small relative to the fiber diameter. This will result in smaller overall deformations (and hence fiber distortions) and more even stress distributions along the lengths of fibers. Second, fibers with low bend stiffness (i.e. small diameter, low modulus) cannot be processed in combination with high strength matrices by techniques known to result in fiber bending (e.g. powder/fiber compacts and spray-deposited monotapes). Third, matrices of lower strength need to be used which will deform more readily and thus keep the fibers unbent and unbroken. Fourth, conditions which allow densification by asperity flattening instead of fiber deflection are favored: (1) higher temperatures, since the matrix flow stress drops more rapidly with increasing temperature than does the fiber stiffness/strength, and (2) lower densification rates, since the asperities' creep strength is quite sensitive to strain rate (power law) while the fibers remain essentially elastic throughout processing.

This study has revealed sometimes significant amounts of fiber fracture during processing. Curtin's strength model as presented earlier (Fig. 2) predicts a composite strength loss of 5-20% (based on the 25-140 fractures/m observed under the range of conditions tested). Indeed, tensile tests of consolidated specimens have indicated strengths which were 15-60% below the predicted rule-of-mixture (ROM) strength [29]. While all of the difference between ROM and measured strengths may not be attributable to process induced fiber fractures, they almost certainly contribute significantly to this difference. The experimental results and modeling insight gained in this study should enable process and material optimization leading to control of fiber damage during consolidation.

6. SUMMARY

An experimental study was conducted to explore the phenomenon of fiber fracture during the hot isostatic and vacuum hot pressing of plasma-sprayed Ti-14wt%Al-21wt%Nb/SCS-6 composite monotape lay-ups. From these results we conclude that:

1. Fiber fracture during the consolidation of plasma sprayed monotape is a significant factor in determining material quality and final properties.

2. The number of fiber fractures during consolidation is strongly dependent upon the processing conditions. The fiber fracture density is directly related to consolidation pressure and inversely related to consolidation temperature. As the rate of pressure application increases, the number of fiber fractures increases ($T < 900^\circ$C). The application of small pressures (3-4 MPa) to unconsolidated composite specimens at room temperature causes fiber fracture (up to 20 fractures/m).

3. The principal mechanism of fiber fracture is bending due to stress concentrations occurring at asperity contacts.

4. The observed dependence of damage on process conditions can be predicted by analysis of a volume element of porous composite in which the interaction between fibers and deforming matrix asperities is captured. The model predicts that increases in fiber stiffness, strength, and diameter, decreases in the matrix yield and creep strength, and decreases in monotape surface roughness will lower the fiber fracture density associated with consolidation.
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REFERENCES


APPENDIX

A micromechanical model relating the macroscopic rate of densification to the local bending of reinforcing fibers is developed. The macroscopic densification rate can be predicted for a laminate of spray-deposited MMC foils given the applied pressure and temperature using a previously developed model [12]. A unit cell assumed to typify the fiber bend geometry (see Fig. 7) is taken to deform uniaxially with the ceramic fiber acting as a simply-supported, cylindrical beam undergoing purely elastic bending up to the point of fracture. The asperities pressing against the fiber are allowed to deform by plastic yielding and by power law creep, these two cases being treated separately to keep the analysis as simple as possible. The asperity contacts are treated as point loads. The maximum tensile stress due to bending can then be used in conjunction with a Weibull description of the fiber strength to calculate the probability of fracture.

From the geometry of the representative unit cell problem in Fig. 7, the total cell height (2z), which is given, must be the sum of the heights of the deformed asperities (2y) and the deflection (v) of the fiber

\[ 2z = 2y + v. \]  

[Note that \( v \) is always negative since \( y > z \) and that initially, \( z = z_0 = y = y_0 \).

Considering first the case of rigid, perfect plasticity, the asperity is assumed to undergo deformation whenever the stress acting at the contact exceeds an effective yield stress

\[ \sigma_z \geq C \cdot \sigma_y, \]  

where \( \sigma_z \) is the matrix material's yield stress in uniaxial tension, and \( C \) is a concentration factor (\( C > 1 \)). This factor accounts for the added effort needed to overcome the elastic constraint of material surrounding the plastic zone, which forms within the interior of spherical bodies in contact [30]. Recent work of Gampala et al. [31] has shown \( C \) to be dependent upon the degree of deformation, but for the present purpose it is sufficient to treat \( C \) as constant. Accordingly, we take \( C = 3 \) corresponding to fully plastic flow during plane strain indentation of a cylindrical punch into a semi-infinite halfspace.

The fiber deflection can be obtained in terms of the input cell height, \( z \), as follows: the contact stress is just \( F/a_z \), where the contact area (\( a_z \)) can be approximated in terms of the asperity displacement as \( a_z \approx 2 \pi z_0 (\gamma + y) \) [30], and the force (\( F \)) acting on the asperity is just the force required to bend the fiber

\[ F = k_s v, \]  

where, from simple beam theory, the fiber bend stiffness, \( k_s \), is

\[ k_s = \frac{3E_s}{4} \frac{d_s^4}{L^3}. \]
Combining (A2) and (A3) and using the approximate expression for the contact area, the deformed asperity height can be expressed as

\[ y = y_0 + \frac{k_s v}{6\sigma_0\sigma_f}. \]  

(A5)

Defining \( k_p \), the plastic stiffness, as \( 6\sigma_0\sigma_f \) and substituting (A5) into (A1) gives the fiber deflection in terms of the imposed cell deformation

\[ v_p = \frac{2(z - z_0)}{1 + 2\cdot\zeta} \]  

(A6)

where \( y_0 = x_0 \) has been used and \( \zeta \) is the dimensionless ratio of fiber bend stiffness to plastic (asperity) stiffness. Deflections are minimized by increasing the value of \( \zeta \), implying high modulus fibers with large diameters (to resist bending) and low matrix yield strength.

At elevated temperatures densification is achieved primarily by means of power law creep. For this time dependent mechanism we are interested, in analogy with (A6), in an expression for the rate of change of fiber deflection as a function of the overall cell deformation rate \( \dot{\varepsilon} \), the fiber bend stiffness \( (k_s) \), and the material parameters characterizing the matrix creep response. Differentiation of (A1) gives for \( \dot{v} \)

\[ \dot{v} = 2z - 2\gamma \dot{y}. \]  

(A7)

The time rate of change in the asperity height (\( \dot{y} \)) can be obtained from the uniaxial power law creep relation

\[ \dot{\varepsilon} = B \cdot \sigma^n \]  

(A8)

where \( B \) is taken to have an Arrhenius dependence on temperature \( [B = B_0\exp(-Q_c/RT_c)] \) and \( n \) is the stress exponent. The strain rate, \( \dot{\varepsilon} \), is rewritten as \( \dot{y}/y \equiv \dot{y}/y_0 \) and the stress is again taken to be the contact stress: \( \sigma = F/a_0 \). Then, taking an average value for the area of contact \( (a_0 \approx 2/3 \cdot \pi y_0^2) \) and equating the force acting on the asperity with that required to deflect the fiber \( (F = k_s v) \), (A8) becomes

\[ \dot{y} = B \left( \frac{k_s v}{a_0} \right)^n. \]  

(A9)

Solving (A9) for \( \dot{y} \) and substituting the result into (A7) gives a nonlinear, ordinary differential equation in the deflection, \( v \)

\[ \dot{v} + 2B_0\left( \frac{k_s}{a_0} \right)^n v^n = 2z. \]  

(A10)

If only constant cell displacement rates are considered and \( n = 2 \), (A10) can be solved analytically to obtain

\[ v_p(t) = \frac{2z}{\sqrt{2\gamma}} \left[ \exp(2t \sqrt{2\gamma}) - 1 \right] \]  

(A11)

where \( \gamma = 2B_0(k_s/a_0)^2 \). Increasing \( \gamma \), (i.e. decreasing the creep resistance of the matrix or increasing the fiber bend stiffness), has the effect of minimizing the deflection. Given the fiber midpoint deflection from either (A6) or (A11), the maximum tensile stress in the fiber \( (\sigma_t) \) can be calculated from

\[ \sigma_t = \frac{16 \cdot F}{\pi d^2}. \]  

(A12)

where \( F \) is determined from (A3). The cumulative probability of a fiber failing is given by a Weibull distribution \( [32] \)

\[ \Phi_\sigma(\sigma) = 1 - \exp \left[ - \left( \frac{\sigma_f}{\sigma_0} \right)^m \right] \]  

(A13)

where \( \sigma_0 \) is a reference stress and \( m \), the Weibull modulus, was found to be 13.0 ± 2.1 (see Table 1 for experimentally measured values).

Since the fiber strength tests were conducted in tension as opposed to bending, the reference stress in the Weibull expression must be related to the stress required to cause fracture in bending. Siemers et al. [33] have derived and experimentally verified such a relationship

\[ \sigma_b = \sigma_t \left[ \frac{V_t}{V_b} \right]^{1/m} \]  

(A14)

where \( \sigma_b \) and \( \sigma_t \) are the bending and tensile stresses, respectively, and \( V_b \) and \( V_t \) are the volumes of fiber exposed to critical stresses in tension and bending, respectively. The factor \( k \) depends only on the Weibull modulus and is 1.45 · 10^{-2} for the SiC fibers considered here. Taking \( V_t \) to be the gage length of tensile fiber samples and \( V_b \) to be the mean asperity spacing, \( V_t/V_b \approx 100 \), gives

\[ \sigma_b \approx 1.9 \cdot \sigma_t. \]  

(A15)

For this range of values, (A14) is quite insensitive to values of \( V_b \), and over the entire range of fiber segment lengths (i.e. the distribution of asperity spacings) encountered, \( \sigma_b/\sigma_t = 1.9 \pm 0.1 \).

\( \dagger \)The solution of this ordinary differential equation \( [i.e. \ y(t)] \) differs by only a few percent from that of equation (16) in Ref. [12] when \( B = C = 3 \).