ANISOTROPIC DAMAGE EVOLUTION IN A 0°/90° LAMINATED CERAMIC-MATRIX COMPOSITE

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Abstract—Anisotropic damage evolution in a 0°/90° laminated Nicalon™ SiC fiber-reinforced calcium aluminosilicate (CAS) glass–ceramic composite during uniaxial tensile deformation has been investigated using a variety of non-invasive characterization techniques. The elastic constant reduction in the three principal directions was measured from in situ laser-generated ultrasonic velocity measurements in various sound propagation directions. They indicate that, in addition to a large drop in elastic stiffness in the loading direction, the constants characterizing the nominal elastic stiffness transverse to the loading direction were also degraded. Surface replicas taken intermittently during loading revealed that transverse softening of the elastic stiffness was associated with fiber/matrix interface damage mainly in the 0° plies, while the large softening of the elastic stiffness in the loading direction was the result of multiple matrix cracking in both the 0° and 90° plies. While the ultrasonic data allowed a detailed characterization of the anisotropic damage evolution in this laminate, acoustic emission measurements and surface replica data identified the crack initiation stress in the 90° plies and correlated it to macroscopically observable deviations of the stress–strain curve from linear elastic behavior. These matrix cracks were found to have initiated preferentially in the weak 90° plies near the 90°/0° ply boundaries. © 2000 Acta Metallurgica Inc. Published by Elsevier Science Ltd. All rights reserved.

Keywords: Acoustic; Non-destructive testing; Multilayers; Ceramics; Composites

1. INTRODUCTION

Laminated composites sometimes have significant advantages over unidirectional composites for multiaxially loaded structural applications because the orientation, thickness and stacking sequence of the fiber-reinforced laminae can be varied to satisfy the needed stiffness and strength requirements in different loading directions [1, 2]. However, ensuring the durability of brittle-matrix laminates can be problematic because damage modes such as transverse ply cracking, fiber/matrix interface debonding and inter ply delamination are more complicated than in the case of a unidirectional composite loaded in one direction. A detailed understanding of the anisotropic aspects of damage evolution is therefore important for both designing improved laminated composites and utilizing them more reliably in structural applications.

When a brittle-matrix 0°/90° cross-ply laminate is loaded with the tensile axis aligned parallel to the 0° ply direction, it is generally believed that transverse matrix cracks first appear in the weaker 90° plies [3]. These small cracks are thought to extend laterally to span the entire 90° layer and only then penetrate into neighboring 0° plies [3–5]. Finite element analyses have indicated that the degradation of the elastic stiffness constants of cross-ply laminates is related to the transverse matrix crack density [1, 5, 6], but relatively little experimental work has been conducted to measure damage evolution in situ and to confirm this viewpoint. Conventionally, damage evolution is characterized by an analysis of the uniaxial stress–strain behavior [3, 7, 8]. In this approach, the effective Young’s modulus in the loading direction at various deformation stages is determined from loading/unloading curves and compared with metallographic observations of the crack density. The recognition of anisotropic damage and the assessment of its mechanical significance have been limited, in part, by the absence of effective experimental methodologies for its measurement.

Recent ultrasonic studies on an SiC/SiC bi-directional composite immersed in a water bath (for ultra-
sonic coupling) indicated that ultrasonic techniques have the potential to characterize anisotropic damage [9]. A laser-ultrasonic (LU) approach has subsequently been developed and used to monitor the anisotropic damage evolution in a unidirectional Nicalon™ SiC fiber-reinforced calcium aluminosilicate (CAS/SiC) composite in situ during uniaxial tensile loading in its fiber direction [10]. This laser-ultrasonic technique revealed that extensive reverse softening of the elastic moduli accompanied degradation of the axial elastic modulus. This was believed to be caused by fiber–matrix debonding near the intersection of fibers with transverse matrix cracks. Here, we apply this LU method to investigate anisotropic damage accumulation in a 0°/90° cross-ply CAS/SiC composite loaded in uniaxial tension.

In this study, anisotropic damage in 0°/90° cross-ply samples was characterized using ultrasonic velocities measured in situ within two principal planes. The anisotropic damage behavior is represented by a deterioration of the elastic stiffness constants $C_{11}$, $C_{22}$, and $C_{13}$ (determined from the ultrasonic velocities) in the three principal directions. In addition, acoustic emission sensing, intermittent surface replica characterizations and loading/unloading hysteresis curve data were acquired. Correlations between the elastic stiffness deterioration and anisotropic damage accumulation are discussed and compared with existing views of damage progression in unidirectional CAS/SiC composites.

2. EXPERIMENTAL APPROACH

Laminated 0°/90° CAS/SiC composite material was provided by Corning, Inc. (Corning, NY). The material was composed of 16 plies with a ply thickness of about 170 μm. The material properties have been presented elsewhere [8, 11]. The dimensions of the tensile specimens were $\sim 150 \text{ mm} \times 10 \text{ mm} \times 2.7 \text{ mm}$. The sample ends were bonded to low-modulus fiberglass tabs for gripping. The edges of the specimens were polished before testing so that acetate replicas could be taken at various stages of loading for crack detection and monitoring of its progression.

The coordinate system used to characterize both the fiber architecture and the test configuration is illustrated in Fig. 1, where, for clarity, only four plies near the middle plane are shown. Direction 1 was the loading direction while 3 was the laminate thickness direction. Continuous SiC fibers were aligned along both the 1- and 2-directions. Since the load was applied in the 1-direction, the plies with fibers in this direction were the 0° plies. Double 90° layers were present at the center of the lay-up, and an even number of 0° and 90° plies existed in the sample. Consequently, the 0°/90° cross-ply used for the tests is a symmetric laminate, and there should be no coupling between stretching and bending [2]. Also shown schematically in Fig. 1 are the directions of laminate thermal residual stresses. Note that each ply is under a biaxial residual stress state.

Tensile testing was performed with a screw-driven Instron 4200 machine at a crosshead speed of 0.03 mm/min. The axial strain was measured with an extensometer (1 in. gauge length). The specimen was loaded to a pre-set stress level, held at that level for laser-ultrasonic measurements, and the stress then reduced to 10 MPa. During the loading and unloading process, acoustic emission events were also recorded. Details of the LU measurement and acoustic emission recording techniques are identical to those reported previously [10].

3. STRESS–STRAIN BEHAVIOR AND VISUAL OBSERVATIONS OF DAMAGE EVOLUTION

3.1. Stress–strain behavior

A typical loading/unloading stress–strain curve for a CAS/SiC cross-ply sample and the corresponding acoustic emission (AE) events are shown in Fig. 2. The deviation of the stress–strain curve from the linear regime occurred at around 50 MPa, and correlated with an acceleration of AE at about the same stress level. A detailed correlation between AE events, the crack initiation stress and the residual stress will be discussed in Section 5.

3.2. Matrix cracking evolution

Figures 3–5 are optical micrographs of surface replicas exhibiting matrix cracks at different stages of loading. At low stress levels (<75 MPa) as shown in Fig. 3(a) and (b), some (marked) matrix cracks in the 90° ply extended across the entire layer; however a majority of the cracks were rather short and confined within the ply boundary by fiber bridgment. The metallography clearly indicated that, in contrast to the viewpoint adopted by many modeling studies, matrix cracks, once initiated, do not always span an entire ply. Instead, the initiation of new matrix cracks and the growth of pre-existing cracks (to span an entire ply) progressed simultaneously. For example, Fig. 4(a)–(c) reveals a series of surface cracks at three different stress levels. A newly initiated crack in the 90° ply at the cross-ply boundary at 93 MPa [marker 1 in Fig. 4(a)] had partially propagated across the ply after the stress was increased to 106 MPa [Fig. 4(b)]. At a higher stress of 120 MPa, this crack had spread across the entire ply [Fig. 4(c)]. Figure 4(c) also shows an example of a matrix crack (marker 2) that was initiated at the high stress level.

Figure 5 presents a higher magnification of a surface replica that shows cracks deflecting around the fibers in the 90° layer leaving a debonded interface. The matrix cracks in the 90° plies exhibited both straight and curved paths, depending on the local arrangement of fibers within the ply. The curved cracks (Fig. 4, marker 1) appeared to have extended more slowly than the straight cracks [see cracks in
the top 90° ply in Fig. 3(b)]. Although previous studies have also indicated that the matrix cracks initiated first in the 90° plies [3, 12], the observations of crack growth presented here are not in total agreement with all of the prior investigations. For example, Beyerle et al. [3] suggested that matrix cracks in the 90° ply always spanned the entire ply once initiated, whereas Mall and Kim [12] believed that matrix cracks occurred in a random manner and that no single crack grew across an entire ply.

The average crack density was determined as a function of stress from the surface replicas and is presented in Fig. 6. Crack initiation stress was taken as 50 MPa, which was estimated from the stress–strain curve and AE measurement. At stress levels below ~95 MPa the crack density in the weaker 90° ply was higher than in the 0° ply. As the applied stress increased, the 90° ply cracks then began to penetrate into the 0° ply. By the time the stress had reached above 95 MPa, matrix cracks in the 0° ply became more prevalent than in the 90° ply because of the overlapping of matrix cracks from the 90° plies on either side of the 0° ply. The penetration of pre-existing cracks into neighboring 0° plies was accompanied by the initiation of new matrix cracks in the 90° ply. This again contrasts with a previous report that matrix cracks in the 90° ply were saturated when the stress had reached 80 MPa [3]. It has also been noted that the ultimate tensile strength of five 0°/90° composite samples tested here ranged from 130 to 160 MPa, while in previous studies the failure stress for this composite was reported from 120 MPa [12] to 220 MPa [3, 13]. These differences may be attributed to sample variations [14].

Metallographic observations of surface replicas taken after unloading and failure are consistent with the notion that matrix cracks initiated in the 90° ply near the ply boundary, then gradually extended into the 90° ply. As shown in Fig. 7 (taken after sample failure), matrix cracks in the central double 90° layer are arrested near ply boundaries. This is believed to be due to uneven crack openings in the 90° ply: crack openings near the ply boundary (initiation site) were generally larger than those at the ply center for a given stress level. Additionally, fibers in the 0° plies exert frictional forces upon crack closure during unloading, which could also contribute to a lower extent of crack closure in the 90° ply near the ply boundary. This is further illustrated by comparing Fig. 8(a), taken at a stress of 137 MPa, with Fig. 8(b) taken after unloading to 10 MPa from 150 MPa. It can be seen that crack openings near the central part of the 90° ply crack (shown by arrow) are significantly reduced after unloading. Since there was no obvious increase in the reloading modulus compared with the unloading modulus, what has been seen in the central region of 90° ply after unloading [Fig.
Fig. 3. Optical micrographs of replicas showing surface cracking in the early stages of deformation at: (a) 60 MPa; (b) 75 MPa.

8(b)] and failure (Fig. 7) is most likely a result of the partial closing of crack opens rather than crack “healing”.

4. LASER-ULTRASONIC CHARACTERIZATION

4.1. Ultrasonic velocity and elastic stiffness

Anisotropic reductions in the elastic stiffness tensor components of damaged ceramic-matrix composites (CMCs) result from the collective effects of microcracks. When the wavelength of sound propagating in a body is much larger than the size of the microcracks it contains, the wave velocities and the elastic stiffness constants, $C_{ij}$, are related through the Christoffel equation [15]. While the symmetry of the as-received material with even numbers of $0^\circ$ and $90^\circ$ plies can be regarded as tetragonal ($C_{11} = C_{22}$), it is reduced to orthotropic after damage in the form of transverse matrix cracks is induced. In the damaged state $C_{11} \neq C_{22}$ and, for such an orthotropic material, the ultrasound velocities of the three different wave modes in the 1–3 plane can be related to the elastic stiffness constants [15].

For the quasi-longitudinal mode (QL), the velocity is given by

$$V_{QL} = \frac{\sqrt{a + \sqrt{b^2 - 4c}}}{2\rho},$$

and for the quasi-shear mode (QT)

$$V_{QT} = \frac{\sqrt{a - \sqrt{b^2 - 4c}}}{2\rho},$$

where
Fig. 4. Optical micrographs of surface replicas taken in the same area at different loading stages: (a) 93 MPa, arrow 1 indicates short matrix cracks initiated at 90° plies near ply boundaries; (b) 106 MPa, showing the crack growth within a 90° ply (arrow 1); (c) 120 MPa, crack 1 spans the 90° ply, arrow 2 shows a newly initiated matrix crack.

\[ a = (C_{11} + C_{55}) \sin^2 \theta + (C_{33} + C_{55}) \cos^2 \theta, \]  
(3)

\[ b = C_{55} + C_{11} \sin^2 \theta + C_{33} \cos^2 \theta \]  
(4)

and

Fig. 5. Optical micrograph showing that the matrix crack propagation (stress 106 MPa) in the 90° layer deflects around fibers; fiber ends in 90° plies are visible due to debonding.

Fig. 6. Average matrix crack densities in 0°/90° plies, 0° and 90° plies as a function of stresses.

Fig. 7. Surface cracks after failure, where cracks in the 90° ply are only visible near the ply boundary.
c = (C_{11} \sin^2 \theta + C_{55} \cos^2 \theta)(C_{55} \sin^2 \theta + C_{33} \cos^2 \theta) - (C_{13} + C_{55})^2 \sin^2 \theta \cos^2 \theta.

For the pure shear mode (PT) we have

\[ V_{pt} = \sqrt{\frac{C_{66} \sin^2 \theta + C_{44} \cos^2 \theta}{\rho}}, \]

where \( \theta \) is the angle between the wave propagation direction and the 3-axis. The expressions for the sound velocities in the 2–3 plane are similar; one only needs to replace the subscript 1 by 2, and 5 by 4.

It is worth noting that the above equations are only valid for homogeneous materials. However, since the ultrasound wavelength is relatively large (>1000 \( \mu m \)) in CAS/SiC compared with the scale of the inhomogeneities involved (fibers with diameter 15 \( \mu m \) and plies with thickness ~170 \( \mu m \)), these CAS/SiC cross-ply composites can be regarded as homogeneous media to a good approximation and the above equations can therefore be applied to convert measured velocity data to elastic stiffness constants [16].

Anisotropic damage along the three principal directions is characterized by \( C_{11}, C_{22} \) and \( C_{33} \). In the process of deducing \( C_{11}, C_{22} \) and \( C_{33} \), a nonlinear curve fitting method, similar to that described previously [10], has been used. An error analysis indicates that the off-diagonal elastic constants have relatively large uncertainties compared with diagonal elastic constants, and so we concentrate on the use of \( C_{11}, C_{22} \) and \( C_{33} \) to study anisotropic damage in the cross-ply laminates.

### 4.2. Elastic constant data in the undamaged state

The elastic properties of undamaged 0°/90° CAS/SiC have been evaluated precisely using resonant ultrasound spectroscopy (RUS) [11]. To compare the results from RUS with LU measurements, the wave propagation velocity in the 1–3 plane was calculated based on elastic constants measured by the RUS technique. These are plotted along with the LU measured wave velocities in Fig. 9. In Fig. 9, the square symbols represent the LU measured longitudinal wave velocities and triangles represent the shear velocities. We note that the RUS predicted and LU measured velocities are in good agreement. In the unloaded states, both the RUS and LU techniques indicate that the PT and QT shear wave velocities in the 1–3 plane are nearly identical.

### 4.3. Degradation of elastic stiffness constants during loading

Figure 10(a) and (b) shows the elastic stiffness constants in the three primary directions determined from the LU velocities as a function of applied stress using a compilation of data from two tests. The Young’s modulus along the loading direction (\( E_1 \)) was determined from the stress–strain curve by partial unloading (\( \Delta \sigma = 20 \text{ MPa} \)) and is also plotted for comparison.

![Fig. 9. A comparison of ultrasonic velocities in the 1–3 plane of the as-received CAS/SiC cross-ply laminate deduced from RUS measurements and the laser-ultrasonic method. Square symbols represent the longitudinal wave mode and triangle symbols represent the shear mode. Experimental data are based on two specimens.](image-url)
Table 1. Elastic stiffness constants of 0°/90° cross-ply CAS/SiC composites determined by RUS [11] and laser-generated ultrasound (units: GPa)

<table>
<thead>
<tr>
<th></th>
<th>$C_{11}$ (±%)</th>
<th>$C_{13}$ (±%)</th>
<th>$C_{15}$ (±%)</th>
<th>$C_{12}$ (±%)</th>
<th>$C_{44}$ (±%)</th>
<th>$C_{66}$ (±%)</th>
</tr>
</thead>
<tbody>
<tr>
<td>RUS</td>
<td>147.45 (0.11)</td>
<td>145.19 (0.28)</td>
<td>52.23 (0.39)</td>
<td>49.96 (0.88)</td>
<td>46.12 (0.01)</td>
<td>46.35 (0.02)</td>
</tr>
<tr>
<td>Ultrasound method</td>
<td>150.0 (3.3)</td>
<td>144.2 (2.4)</td>
<td>54.2 (14.6)</td>
<td>–</td>
<td>42.5 (10.1)</td>
<td>45.3 (4.9)</td>
</tr>
</tbody>
</table>

Fig. 10. (a) Elastic constants $C_{11}$, $C_{22}$ and $C_{33}$ determined from laser-ultrasonic wave velocity measurements. Young’s modulus $E_1$ measured from unloading stress–strain curve is also shown. (b) Normalized curves showing the relative trends of stiffness reduction.

Normalized data are also included [Fig. 10(b)] to show the relative reductions. Similar to unidirectional CAS/SiC, $C_{11}$ in the loading direction exhibited the largest stiffness reduction as a result of many transverse matrix cracks. Its relative trend with stress [Fig. 10(b)] is about identical to the change of $E_1$ measured by mechanical testing. $C_{11}$ and $C_{22}$ have approximately the same initial value, but lose their degeneracy as damage develops. Normalized data indicate that $C_{22}$ and $C_{33}$ have a similar reduction trend [Fig. 10(b)]. The softening of $C_{22}$ and $C_{33}$ implies that cracks with opening displacement in the 2- and 3-direction exist and are most likely a consequence of fiber/matrix interface debonding.

Although debonding occurs both in the 0° and 90° plies, softening in the transverse plane is presumed to be a consequence of interfacial debonding in the 0° plies. This is because the matrix cracks are nearly parallel to the transverse plane, and micromechanical calculations show that the elastic constants in the directions parallel to the crack plane are insensitive to the crack density [17]. Because 0°/90° laminates have only half the number of 0° plies compared with unidirectional laminates, the stiffness reduction in the transverse direction for the 0°/90° laminates is therefore less than that observed in unidirectional CAS/SiC [10, 18].

Figure 8 indicates that partial crack closure appears to have occurred in the 90° ply upon unloading. This is consistent with the observation of a slight increase in ultrasonic velocity during unloading to 10 MPa. The deduced values of $C_{11}$ and $C_{15}$ during loading and unloading are shown in Fig. 11. Error bars (similar to those given in Fig. 10) are not included for clarity. Since partial crack closure occurs mainly in the loading direction, an increase in $C_{11}$ is seen during unloading while $C_{33}$ appears to have remained constant. It should be pointed out that the results in Fig. 11 are from two tests, and some of the scatter can be attributed to sample variations.

5. RESIDUAL STRESS AND MATRIX CRACK INITIATION

The initiation stress for matrix cracking is an important issue for fiber-reinforced CMCs, because it denotes not only the onset of damage but also the loss of protection provided by the matrix against environmental corrosion and/or oxidation of the fibers [19]. Prediction of the matrix crack initiation stress ($\sigma_{mc}$) requires knowledge of the (statistical flaw population controlled) strength of the ceramic matrix and an understanding of the residual stress state. There are two main components to the residual stress in laminates: that of the individual lamina, and that within...
an isolated ply caused by the thermal mismatch between the fiber and the matrix [4].

The first part (ply residual stress) arises because of the anisotropy in the coefficients of thermal expansion (CTE). In CAS/SiC, the CTE transverse to the fiber direction \((\alpha_t = 4.5 \times 10^{-6} \text{ K}^{-1})\) is larger than in the fiber direction \((\alpha_c = 4.3 \times 10^{-6} \text{ K}^{-1})\) [20]. Clearly, if \(\alpha_c = \alpha_t\), there will be no thermal residual stress resulting from differently oriented plies. Using lamination theory [2] for a symmetric laminate (the case in our study), the residual stress from this source is uniform within each layer, but changes discontinuously across the ply boundary.

Following lamination theory and assuming that each layer is homogeneous [2], the lamination residual stress in each ply is calculated to be biaxial with a magnitude of about 13 MPa. The 90° plies are under tension in the loading (i.e., global 1-) direction while the 0° plies are in compression in the 1-direction, Fig. 1. Because the residual stresses in the 0° and 90° plies are of identical magnitude but opposite in sign, the net residual force on the laminate is zero.

To this stress state we add the effects due to the CTE mismatch between the fiber and the matrix. Since the matrix has a higher CTE than the fiber, the matrix is under axial and circumferential (hoop) tension, and the fiber/matrix interface (radial direction) is under compression. Using an elastic composite cylinder model [21, 22], the calculated residual matrix axial stress \(\sigma_{m,ax}^R = 81\) MPa and the radial interfacial stress \(\sigma_{m,rr}^R = -57\) MPa, assuming 0.35 fiber volume fraction. These values are in reasonable agreement with previously reported values (axial 89 MPa and radial -65 MPa) [8]. The hoop stress, \(\sigma_{m,rr}^R\), in the matrix calculated from the composite cylinder model is tensile with a maximum at the fiber/matrix boundary of ~120 MPa. For the laminate system investigated here, the contribution from the laminae residual stress is much less than the residual stress resulting from CTE mismatch between the fiber and the matrix.

When an external stress \(\sigma_{11}\) is applied, the stress is distributed among the layers. Using lamination theory, the stress distributions on 0° and 90° layers are approximately 1.035\(\sigma_{11}\) and 0.965\(\sigma_{11}\). Thus, the maximum total stress along the loading direction, including contributions from applied load, lamina residual stress and hoop residual stress, on the 90° ply prior to matrix crack initiation is

\[
\sigma_m = 0.965\sigma_{11} + \sigma_{00}^R + \sigma_{90}^R. \tag{7}
\]

For unidirectional CAS/SiC, the total stress carried by the matrix is

\[
\sigma_m = \frac{E_m}{E_c} \sigma_{11} + \sigma_{00} = 0.76\sigma_{11} + \sigma_{00}. \tag{8}
\]

Surface matrix cracks are detected when the applied stress \(\sigma_{11}\) reaches 140 MPa for unidirectional CAS/SiC [8, 10], which corresponds to \(\sigma_m = 187.4\) MPa from equation (8). When the maximum stress criterion is applied, based on the matrix crack initiation stress of the unidirectional composite, the predicted applied stress for matrix crack initiation in the 0°/90° cross-ply laminate is 56 MPa [equation (7)]. This is in good agreement with the lowest stress level where surface cracks were visually detected [Fig. 3(a)]. Mechanical testing shows that the stress-strain curve starts to deviate from linear behavior at ~50 MPa, and the corresponding AE recording also indicates that the onset of matrix cracking occurs at about the same stress level (Fig. 2). Other work has also shown that the matrix cracking stress \(\sigma_{11}\) ranges from 40 to 60 MPa [3]. Some of this discrepancy can be attributed to sample differences. Additionally, it has also been noted that in unidirectional CMCs, matrix cracking initiation depends on the local fiber distribution; matrix cracking can be initiated in the matrix-rich region at a lower stress level than where the fibers are more densely and uniformly distributed [23]. Also, free edge effects (although considered to be small in this case) on surface crack initiation in a 0°/90° laminate [24] were not investigated in above simplified analysis. Regardless, the overall agreement between theory and experimental observation is good.

Finally, we note that the residual stress analysis above showed that the maximum matrix tensile stress occurred in the loading direction, within the 90° ply. The observation that matrix cracks initiated preferentially near the boundaries of 90° plies must be either a consequence of a locally higher matrix fraction at the ply interface or variations in stress at the 0°/90° ply boundary [25]. Further studies are needed to improve the incorporation of matrix crack initiation in models of laminate mechanics.

### 6. CONCLUSIONS

The elastic stiffness degradation of a 0°/90° cross-ply laminate has been evaluated by a combination of a laser-ultrasonic characterization, acoustic emission and surface replicas. The LU technique allowed a detailed characterization of the anisotropic damage through ultrasonic velocity measurements in various propagation directions during loading and unloading of the composite. Damage along the three principal directions was characterized by the elastic constants \(C_{11}, C_{22}\) and \(C_{33}\). The largest stiffness reduction was observed in \(C_{11}\), the loading-direction modulus. This has been linked to transverse matrix cracking initiating in 90° plies with an initiation stress of about 50 MPa. Smaller reductions for \(C_{22}\) and \(C_{33}\) in the transverse plane are attributed to interface damage in the 0° plies that initiated at about 75 MPa. Matrix cracks in this cross-ply composite are found to initiate at much lower applied stress levels than in unidirectional CAS/SiC composites. This observation is rationalized by residual stress analysis.
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