Life Prediction Methodology for Titanium Matrix Composites

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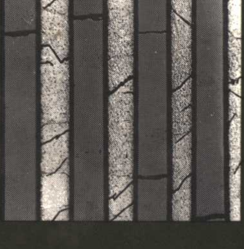

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The quality of the papers in this publication reflects not only the obvious efforts of the authors and the technical editor(s), but also the work of these peer reviewers. The ASTM Committee on Publications acknowledges with appreciation their dedication and contribution to time and effort on behalf of ASTM.

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Foreword

The papers in the publication, *Life Prediction Methodology for Titanium Matrix Composites,* were presented at a symposium held 22–24 March 1994 in Hilton Head Island, South Carolina. The symposium was sponsored by ASTM Committee D30 on High Modulus Fibers and Their Composites, E8 on Fatigue and Fracture Mechanics, and NASA Langley Research Center. W. S. Johnson, Georgia Institute of Technology, J. M. Larsen, USAF Wright Laboratories, and B. N. Cox, Rockwell International Science Center, presided as symposium cochairmen and are coeditors of this publication.

Contents

INELASTIC MATERIAL BEHAVIOR AND MODELING

FATIGUE

Overview

On 22–24 March 1994, a symposium entitled *Life Prediction Methodology for Titanium Matrix Composites* was held on Hilton Head Island, South Carolina. The symposium was sponsored by ASTM Committees D30 on High Modulus Fibers and Their Composites, E8 on Fatigue and Fracture Mechanics, and NASA Langley Research Center. Thirty-three papers were presented by researchers from government laboratories, universities, and industry. The volume represents a collection of those papers that were submitted and passed peer review by three knowledgeable referees.

The rationale for this symposium was to collect the current state of the art in life prediction methodology for Titanium Matrix Composites, TMCs. From the mid 1980's through the early 1990's several large technical programs pushed the development of TMCs and the development of the related life prediction methodology. Most visible among these programs were the National Aerospace Plane (NASP), the Integrated High Performance Turbine Engine Technology Program (IHPTET), and the Hi-Temp Program. These were large programs sponsored by various branches of the U.S. government. The NASP emphasized TMC for structural airframe applications on hypersonic vehicles, while the IHPTET (USAF) and Hi-Temp (NASA) programs developed TMCs for improved gas turbine engine performance. Of course researchers outside of the United States were also involved in TMC studies.

In order to develop accurate life prediction methodologies, a very through understanding of the mechanical damage mechanisms and environmental effects must be established. The envisioned usage conditions of these TMC materials are extreme. Temperature variations between 816°C and −130°C may be encountered on a hypersonic vehicle structure. Cyclic loads up to 60% of ultimate strength would be encountered routinely. Under these conditions the matrix material may not only undergo time dependent deformations but may undergo metallurgical changes as well. Many of the systems discussed in this volume are metastable beta-alloys of titanium. These materials may age with time at temperature in such a way that their mechanical properties, such as ultimate strength, yield strength, and modulus may increase over time. However, at elevated temperatures these materials become sensitive to oxidation, which causes embrittlement and enhanced susceptibility to cracking. Further complications arise because these composites that have fibers and matrices with significantly different coefficients of thermal expansions. Thus as temperature changes so does the internal stress state in the fibers and matrix. Because of this strong influence of thermally induced stresses, the approaches developed for thermal mechanical fatigue of metal that deal with the determination of mechanically induced stresses and strains are no longer applicable. Mathematical analysis must be developed to predict these complex internal stresses since they cannot be measured directly. Thus, the thermo-viscoplastic behavior of the constituents must be determined. Since both the fiber and matrix are very strong, the fiber/matrix interface is usually the weak link in the progression of mechanical failure. Understanding how the strength of the interface can change with time and how the interfacial strength is effected by the thermal residual stresses presents unique challenges on the microstructural scale. The interfacial strength play a significant role in the overall strength and fatigue resistance of TMCs.

This Special Technical Publication is organized into five sections that progress from the basic understanding to the life prediction methodology. The sections are as follows:

Interfacial Properties and Microstructure—Five papers are directed toward understanding the interplay between interfacial strength, residual thermal stresses and the effects of time at temperature.

Fiber Bridging Behavior—Four papers focus on the important role of fiber bridging in the progression of damage. Some papers present modeling approaches while others make detailed experimental observations of the bridging process.

Inelastic Material Behavior and Modeling—Seven papers deal with the complex aspects of experimentally determining inelastic material responses and how to model such behavior in a composite.

Fatigue—Seven papers discuss the various aspects of fatigue damage evolution.

Life Predictions—Six papers present various approaches to tying together all the basic understanding into a methodology that will allow one to predict total life of a TMC.

The collection of work presented here represents tremendous progress toward the characterization and modeling of Titanium Matrix Composites. Nearly all the work was conducted over a relatively short period of time, beginning in 1987. This body of understanding shows that TMCs can be attractive structural materials in stiffness and strength driven designs at elevated temperatures. Like any material, TMCs have their limitations in terms of usage temperature and life. In spite of their high costs, TMCs should find applications that demand their unique blend of mechanical properties and temperature tolerance.

We, the editors of this volume, would like to acknowledge the valuable support of our Steering Committee: Prof. Paul Bowen, University of Birmingham; Dr. Dave Buchanan, McDonnell-Douglas Aircraft; Dr. Rod Ellis, NASA Lewis Research Center; Prof. Tony Evans, Harvard University; and Dr. Ted Nicholas, USAF Wright Laboratory. Further we would like to thank the numerous peer reviewers that collectively spent hundreds of hours to ensure that each paper published in this volume is of a high quality. Lastly, our appreciation is extended to the ASTM staff for all their help in organizing the symposium, collecting the papers, assisting in the review process and publishing this volume.

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Interface Properties and Microstructure

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Interfacial Mechanics and Macroscopic Failure in Titanium-Based Composites

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ABSTRACT: Results are presented for Ti-6Al-4V containing silicon carbide (SiC) monofilaments having duplex (carbon plus TiB_2) coatings. Single fiber pushout testing has been used to study the debonding and frictional sliding characteristics under shear loading, with and without the residual radial compressive stress being reduced by applied in-plane tension. It is shown by finite element method (FEM) modeling that it is important to take account of thermal residual stresses when interpreting data from these tests. Prior heat treatments causing interfacial reaction tend to raise the resistance of the interface to debonding and sliding. This is correlated with data from tension testing of composites under axial and transverse loading, with continuous Poisson's ratio monitoring. The behavior under transverse loading is particularly sensitive to the mechanical response of the interface. The presence of interfacial reaction layers tends to inhibit the development of interfacial debonding and void formation under transverse loading, but causes embrittlement and leads to a reduced strain to failure.

The behavior under transverse load has also been studied with superimposed thermal cycling between 400 and 700°C. Strain histories have been monitored using scanning laser extensometry. Under these conditions, thermal stresses have a pronounced influence on the behavior. Many of the steady-state creep characteristics can be successfully modeled on the basis of matrix creep being controlled by volume-averaged stresses (predicted using the Eshelby method), with stress relaxation processes simulated via a variable stress-free temperature. However, the behavior tends to be influenced from a relatively early stage by interfacial damage development. Interfacial debonding and damage, promoted by a combination of opening mode stress from the applied load and shear stress from differential thermal contraction (particularly towards the specimen edges), soon starts to influence the strain history and has a strong influence on the rupture strain. In sharp contrast to the room temperature behavior, the heavily reacted specimens exhibited delayed onset of interfacial damage and a much longer lifetime, although the final strains to failure were similar. These results are considered in terms of stress fields and interfacial properties.

KEYWORDS: titanium, titanium matrix composites, life prediction, titanium alloys, fatigue (materials), modeling, thermal cycling, creep (materials)

It is now clear that the nature of the interface is of critical importance in controling the behavior of silicon carbide (SiC) monofilament-reinforced titanium, particularly under transverse loading. There have been several studies [*1–10*] of the inelastic deformation and failure characteristics of this class of composites under tensile loading. Several of these studies [*1,2,5–9*] have involved Poisson's ratio measurement during straining as a means of

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monitoring the inelastic processes taking place. The systems studied include Ti-15V-3Al-3Cr-3Sn (fully β) with Textron SCS-6 (carbon core, 3 μm carbon-rich surface layer) fibers [*2–6,8–10*], Ti-6Al-4V (α/β) with SCS-6 fibers [*1,3,4,10*] and Ti-6Al-4V with BP Sigma (tungsten core, 1 μm carbon + 1 μm TiB_2 layer) fibers [*7*]. In all of these cases, axial loading leads to matrix plasticity before final failure, which occurs soon after the fibers start to fracture. Under transverse loading, however, interfacial damage tends to dominate the behavior. This apparently occurs at relatively low macroscopic strains, particularly in locations where the applied load generates a strong Mode I component, but this does not immediately lead to macroscopic failure. Regimes have been identified in which interfacial debonding, local matrix plasticity, and global matrix plasticity occur. A general conclusion [*1,2,9*] is that the interface is inherently weak, but interfacial contact is maintained initially by the radial compressive stress resulting from differential thermal contraction.

In view of the significance of interfacial effects, particularly under transverse loading, several attempts [*1,7,10,11*] have been made to establish correlations between macroscopic properties and directly measured mechanical characteristics of the interface. (There is scope for altering the interfacial properties, not only by changing the fiber coating, but also by allowing various degrees of interfacial reaction to occur during fabrication and subsequent heat treatments.) However, very few such correlations have been successfully established. This is mainly because of limitations and interpretation difficulties associated with most interfacial test procedures. The one most easily applied to these materials is the single fiber push-out, or push-down, test [*12–18*]. Fiber pushout data generally indicate [*1,19–21*] that the interfacial shear strength (and frictional sliding stress) is raised substantially as interfacial reaction proceeds and the graphitic layer is consumed. The associated changes in the transverse tensile strength of the composite, on the other hand, appear to be small [*1,7*]. However, recent finite element method (FEM) studies [*22–24*] of the push-out test have confirmed that (*a*) the interface is loaded predominantly in shear (whereas the opening mode response probably controls the onset of damage in transverse tension testing), (*b*) thermal residual stresses have a very strong influence, and (*c*) loading conditions vary considerably along the length of the fiber. The recently-proposed [*24,25*] tensioned push-out test can in principle be used to explore the effect of Mode I loading, but it has not yet been systematically applied to these materials.

While the room temperature behavior of titanium-based metal-matrix composites (Ti MCCs) is important, interest really centers on the performance at high temperatures. A prime motivation for the incorporation of fibers lies in the dramatic improvement produced in creep resistance when loaded along the fiber direction. However, the creep behavior under transverse loading is of concern. While relatively little has been published [*26–29*] on creep of Ti MMCs, it is known that transverse creep rates can be relatively high, particularly if thermal cycling is imposed. Such cycling has little effect on the behavior when loaded along the fiber direction, but causes marked enhancement of creep under transverse loading [*29,30*]. This is consistent with a progressive transfer of load to the fibers under axial loading, but with matrix creep continuing in a quasi steady state for the transverse case.

It is anticipated that the variations in matrix stress state as the temperature is changed will affect the transverse creep behavior. Less attention has been paid to the fact that interfacial stress states will also be changing dramatically and that this is expected to affect the onset and development of damage processes and hence the creep rupture characteristics. In the present paper, an attempt is made to rationalize data from interfacial and macroscopic room temperature tests, together with results from isothermal and thermal cycling creep experiments.

Interfacial Testing

Material

Composite material was prepared at British Petroleum (BP) by vacuum hot-pressing stacks of aligned SiC monofilaments (coated with carbon/TiB_2-duplex layers, each about 1 μm thick) interleaved between 75-μm-thick Ti-6Al-4V foils, to produce unidirectional composite panels containing about 30% by volume of fibers. Post-fabrication heat treatments were carried out in sealed silica ampoules evacuated to about 10^{-5} torr. In view of the danger of residual oxygen penetrating along fiber/matrix interfaces [*31*], final cutting and machining operations (removing any material contaminated in this way) were carried out after the heat treatment. Composites were heat treated at 865°C for times ranging from 3 to 26 h. It has been confirmed [*7*] that the graphitic layer is progressively consumed at this temperature and has disappeared after the longest of the times employed. During this process, needles of titanium boride (TiB) grow from the coating into the matrix.

Single Fiber Push-out Testing

Wedge-shaped specimens were prepared from composite panels as described in previous publications [*7,21,32*]. Push-out testing was carried out on a number of fibers, the push-out load, and distance from the end of the wedge (and hence fiber aspect ratio = length/diameter) being noted for each fiber. Results from these tests are shown in Fig. 1 in the form of push-out stress as a function of aspect ratio for three heat treatments. The interfacial shear strength values shown for each specimen were obtained from the gradients of the push-out stress/aspect ratio plots, assuming the shear stress to be approximately constant in each case along the length of the fiber during the test. It can be seen that the shear strength rises as reaction proceeds, with a sharp rise apparently accompanying the final disappearance of the graphitic layer (between the 13- and 26-h heat treatments).

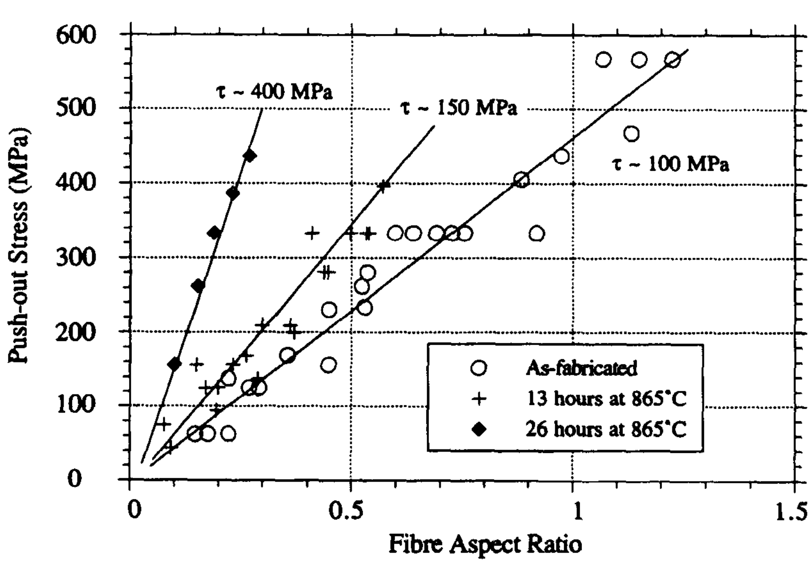

FIG. 1—*Single fiber push-out data* [7] *for specimens with different prior heat treatments.*

In fact, it has become clear from FEM calculations that the assumption of a uniform shear stress being generated along the length of the fiber during push-out testing is a poor one, even for short fibers. This is illustrated by the predictions shown in Fig. 2, giving distributions of interfacial shear stress under typical conditions, with and without thermal residual stress. (The value of 500 K is broadly consistent with the limited data [*33*] from X-ray measurement of thermal stresses.)

Clearly, the thermal stresses have a strong effect. Large variations in shear stress along the fiber length are generated, with the peak, where debonding should initiate, being located close to the bottom of the fiber. (This is predicted over a wide range of values for the applied axial compressive stress, σ_{zA}.) Such features should obviously be borne in mind, but they do not invalidate the deduction from experimental data that the interfacial shear strength rises as reaction proceeds. A further point worthy of note is that a large compressive radial stress also arises from differential thermal contraction (see following paragraphs). This confirms that all such testing normally yields information about how the interface responds to essentially pure Mode II conditions, with no opening mode loading.

Tensioned Push-Out Testing

Tensioned push-out testing has been carried out on composites in the as-fabricated condition, using the preparation procedures and operational details outlined previously [*25*]. A specimen of uniform thickness is required and the push-out load is measured with simultaneous applied in-plane tension (that is, normal stresses transverse to the fiber axis). It has been confirmed by FEM calculations [*24*] that the imposition of these tensile stresses (*a*) has little effect on the interfacial shear stresses and (*b*) strongly affects the interfacial radial

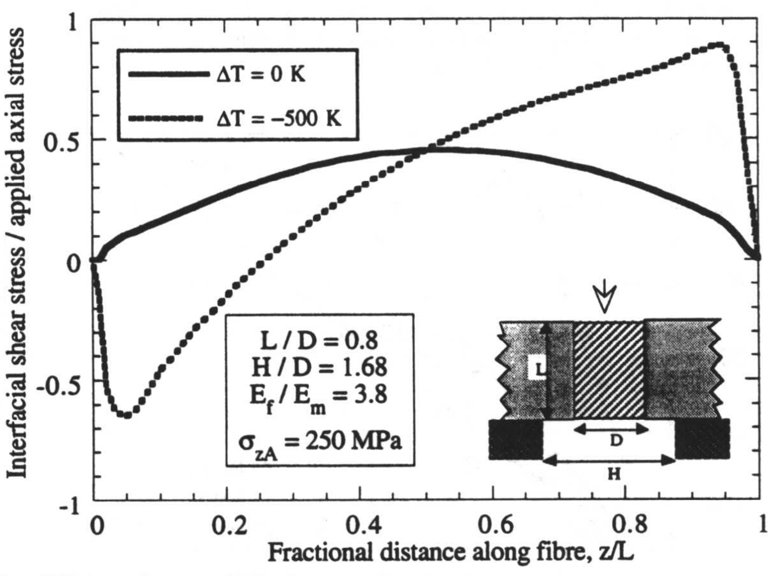

FIG. 2—*FEM predictions* [24] *showing the distribution of interfacial shear stress as a function of distance below the indenter, for typical conditions during push-out of SiC fibers in titanium, with and without thermal residual stress corresponding to an effective temperature decrease during fabrication of 500 K. The stress,* σ_{zA}, *is the applied normal compressive stress along the fiber* (z) *axis.*

stresses. The latter point is illustrated by Fig. 3 that shows for a typical case how the compressive radial stress from thermal contraction can be largely offset by the applied tension. It follows that this test can be used to examine the response of interfaces to loading states covering a range of mode mixities.

An example is shown in Fig. 4 of data obtained with the tensioned push-out test and interpreted using FEM modeling. This shows push-out load values obtained with a single specimen in the as-fabricated condition, converted to an average interfacial shear stress along the fiber length and plotted against the average interfacial radial stress. These data confirm that the application of an opening mode stress increases the ease of interfacial debonding/sliding, as expected. Extrapolation of the data to zero radial stress suggests that the interface will debond spontaneously at that point, with no shear stress or positive tension being necessary. This remains to be confirmed, although several workers [*1,2,9*] have pointed out that observations from macroscopic tests suggest that the interface in these composites (when a graphitic layer is present) seems to debond as soon as the residual compressive stress is offset by the applied load. Further work still required also includes the application of this test to reacted material in which the graphitic layer has been reduced in thickness, or eliminated.

Tensile Properties at Room Temperature

Specimen Preparation

Coupons for tension testing were machined from 1-mm-thick (6-ply) or 3-mm-thick (18-ply) unidirectional composite panels, using an electric-discharge machining technique to produce specimens with gage dimensions of 19 by 5 mm. The gage length was either parallel or transverse to the fiber axis. Gripping was with self-tightening friction grips and specimens were loaded to failure on a screw driven Schenk testing machine with a strain

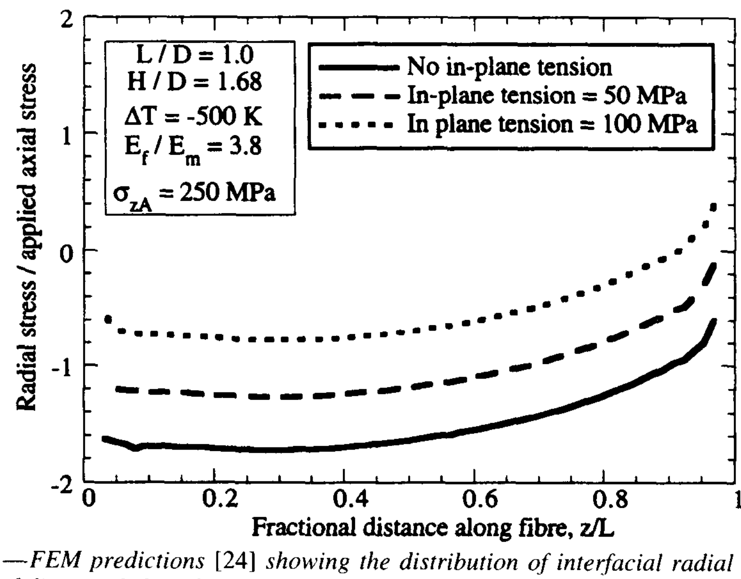

FIG. 3—*FEM predictions* [24] *showing the distribution of interfacial radial stress as a function of distance below the indenter, for typical conditions during push-out of SiC fibe.·s in titanium, with and without applied in-plane tensile stresses.*

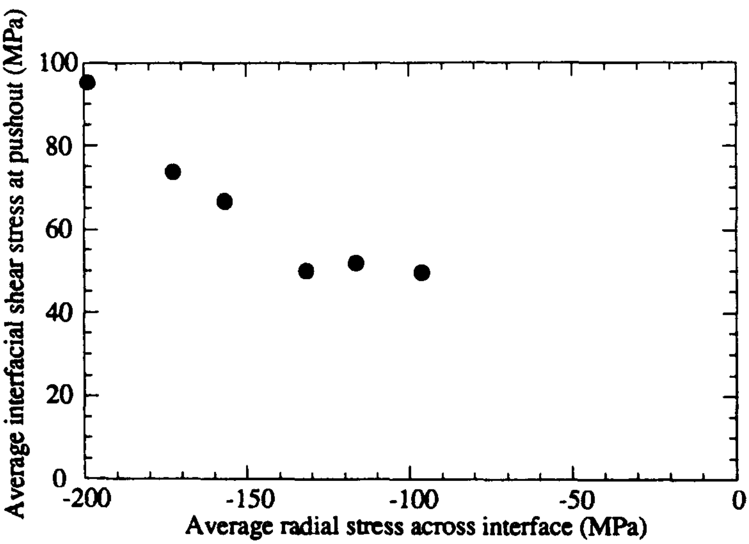

FIG. 4—*Experimental data* [25] *from tensioned push-out testing, plotted in the form of the average interfacial shear stress during frictional sliding as a function of the net average radial stress across the interface, taking account of thermal stresses* [24].

rate of 1.8 10^{-4} s^{-1}. Strain gages were applied to the specimen within the gage length. These were used to monitor strains in loading and longer transverse directions and, in some cases, in all three principal directions.

As-Fabricated Material

Data from axial loading of the composite are shown in Fig. 5. In the initial elastic regime, the experimental Young's modulus, E_1, and Poisson's ratio, ν_{12}, are close to the values predicted theoretically, using the Eshelby method [27] or simpler analytical expressions [34]. The inelastic behavior that is subsequently exhibited is consistent with matrix plasticity being the dominant feature. For example, the observed rise in Poisson's ratio is consistent with this.

Results from transverse loading experiments are presented in Fig. 6. Again the measured elastic constants in the early parts of the curves are broadly consistent with predicted [27] values. Beyond this region, however, the behavior is more complex. A sharp fall initially occurs in the Poisson's ratios, particularly in ν_{23}. This is attributed to interfacial debonding in regions where the fiber radial direction is parallel to the applied load that allows increments of extension in the loading direction without the corresponding lateral contractions (since matrix deformation is shielded by the fibers in this locality). This causes a fall in the Poisson's ratios. Subsequently, local and global matrix plasticity occur, probably while further debonding and voiding (gap formation) are occurring in interfacial regions. This is accompanied by a sharp rise in ν_{23}, while ν_{21} falls away to virtually zero. This reflects the fact that plastic flow is occurring under plane strain conditions, with all of the straining taking place within the 2–3 plane. That such behavior is expected can readily be demonstrated by examining volume-averaged matrix stresses in the elastic regime, again conveniently estimated using the Eshelby method. Predictions are presented in Fig. 7 that clearly

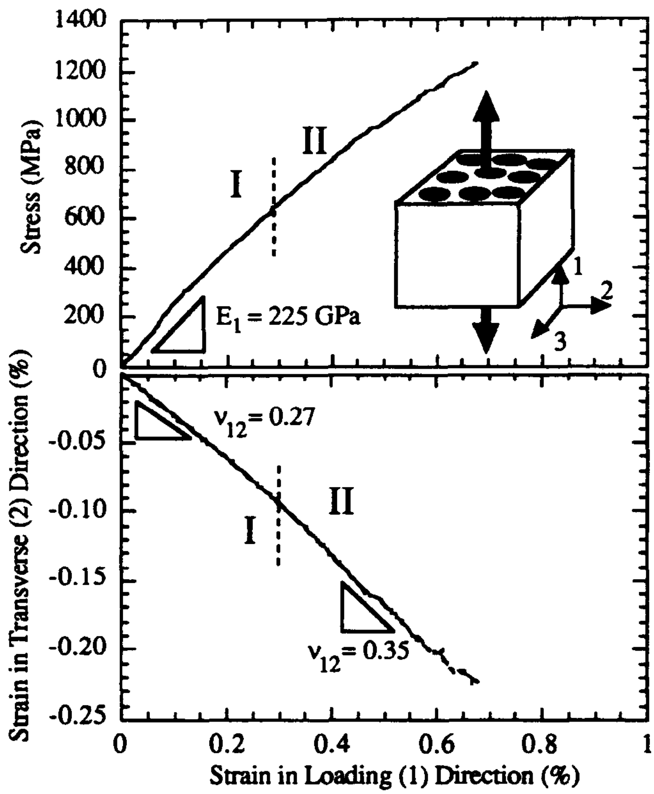

FIG. 5—*Experimental data from axial loading of the as-fabricated composite, up to final failure. The data indicate a transition from elastic behavior (Stage I) to a regime of matrix plasticity (Stage II). Predicted elastic values of the Young's modulus,* E_1, *and Poisson's ratio,* ν_{12}, *are as indicated (numerically and by means of the triangle gradients). The experimental value of* ν_{12} *in the plastic regime is also shown.*

show that the largest deviatoric stresses are in the 2–3 plane, provided the applied stress is greater than about 50 MPa.

The observation that interfacial debonding appears to start at an applied stress of around 170 MPa can be used to explore the local conditions under which this is stimulated, again using Eshelby calculations. The corresponding stress in the fibers along the loading direction is about 200 MPa, and this must also be the peak interfacial radial stress along the loading direction. The compressive radial stress at the interface from differential thermal contraction, on the other hand, is calculated to be about 150 MPa (for a 500 K temperature drop). These estimates are obviously approximate, but they do suggest that debonding occurs in opening mode soon after the net radial stress becomes tensile. More accurate analysis requires FEM modeling (and a reliable estimate of the appropriate temperature drop to employ). An idea of the extent of the region in which debonding is expected to occur can be obtained from the FEM stress field shown in Fig. 8. It can be seen that debonding is initially expected over only a small proportion of the fiber circumference, although the debonded region may grow as the applied load is increased. The details depend on the arrangement of fibers.