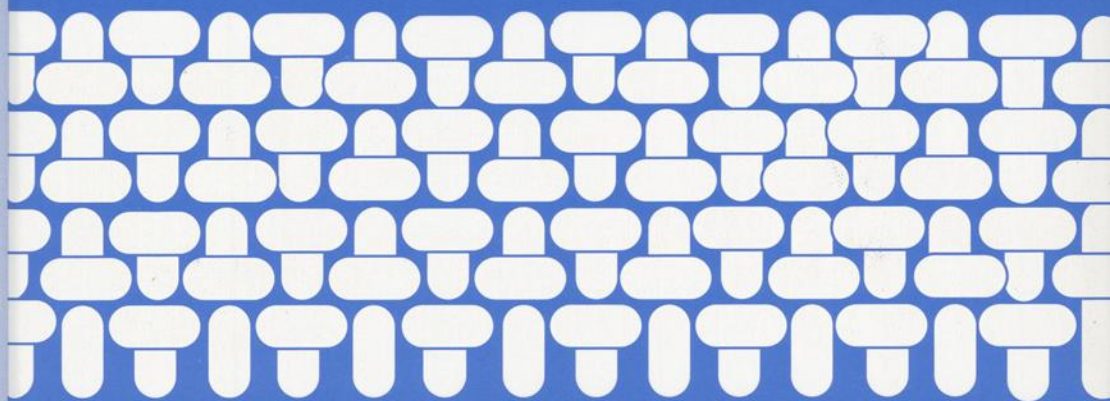


COMPOSITE MATERIALS



Testing and Design

Tenth Volume
Glenn C. Grimes, editor

 **STP
1120**

STP 1120

Composite Materials: Testing and Design (Tenth Volume)

Glenn C. Grimes, editor

ASTM Publication Code Number (PCN)
04-011200-33



ASTM
1916 Race Street
Philadelphia, PA 19103

ISBN: 0-8031-1426-5

ASTM Publication Code Number (PCN): 04-011200-33

ISSN: 0899-1308

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Each paper published in this volume was evaluated by three peer reviewers. The authors addressed all of the reviewers' comments to the satisfaction of both the technical editor(s) and the ASTM Committee on Publications.

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.

Printed in Baltimore, MD
January 1992

Foreword

The Tenth Conference on Composite Materials: Testing and Design was held 24–25 April 1990 in San Francisco, California. ASTM Committee D-30 on High Modulus Fibers and Their Composites sponsored the conference. Glenn C. Grimes, Lockheed Advanced Development Company (The Skunk Works), served as conference chairman and editor of this publication. Most of the papers presented are in this volume, which complements the first through the ninth conference publications with the ASTM STP numbers of 460, 497, 546, 617, 674, 787, 893, 972, and 1059, respectively, all titled *Composite Materials: Testing and Design*. The first Conference was held in 1969 in New Orleans, Louisiana, and has been held approximately every two years since then.

Acknowledgment

As the Symposium Editor, I wish to express my thanks for the effort put in by the eight Session Chairmen in conducting their sessions in San Francisco, California, getting presentation and technical paper reviewers, mediating differences between authors and reviewers, and writing a short summary of the technical papers in their session. I want to give special thanks to Paul Lagace (Session I), Michael Castelli (Session II), Stephen Swanson (Session III), Robert Croman (Session IV), Tom Dunyak (Session V), Karl Steiner (Session VI), Kevin O'Brien (Session VII), and Susan Avery (Session VIII) for their assistance in the accomplishment of this task.

In addition, I must express my appreciation for the approval and support of my endeavor by Robert Goetz, Director of Engineering, and Ben Rich and Sherman Mullin, the immediate past and current Presidents of Lockheed Advanced Development Company (The Skunk Works).

Finally, the help of my secretary, Marylyn Dahlquist, for all the word processing and mailings and the help of my engineering assistant, Edmond Dusablon, for all his telephoning, proofreading, and critical comments are recognized with my sincere appreciation.

Glenn C. Grimes

Chairman and Editor,
10th Symposium on Composite Materials:
Testing and Design

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Overview

Introduction

The ASTM Tenth Conference on Composite Materials: Testing and Design is held so experts in the field of composite materials and structures from government, industry, and academia can present, discuss, and publish the current research and development on composites and the state-of-the-art engineering applications to structures. Important composite materials technical issues like test methodology, material properties and failure criteria, advanced materials research and development, structural analysis and test verification, quality assurance, interlaminar fracture, and damage and repair are discussed.

The general subject matter and purpose of the Tenth Conference was similar to that of the First Conference that was held 21 years ago and chaired by S. Yurenka Elliott, even though the progress in composites technology since has been great. J. Whitney, who chaired the Seventh Conference, discussed this progress. In the First and Second Conferences chaired by H. T. Corten, considerable emphasis was placed on test methodology. It should also be noted that design and fracture were covered in the first two Conferences. Fatigue, creep, and moisture effects received much attention in the Third Conference chaired by C. A. Berg, F. J. McGarry, and S. Y. Elliott. Fracture and impact response was supplemented by environmental effects in the Fourth Conference chaired by J. G. Davis, Jr., whereas, the Fifth Conference chaired by S. W. Tsai covered a wide range of effects including fatigue, time-dependent behavior, and nondestructive evaluation (NDE). The Sixth Conference, chaired by I. M. Daniel, discussed toughness and damage tolerance, failure mechanics, delaminations, and strength. The Seventh Conference, chaired by J. Whitney, included the technologies of structures, failure mechanics, strength, delamination, and analysis and characterization. The Eighth Conference, chaired by J. D. Whitcomb, covered analysis, impact and compression testing, materials characterization, failure mechanisms, nondestructive evaluation, and filament wound and woven composites. Structural considerations and analysis, delamination initiation and growth, damage mechanisms and test procedures, and general interest design analysis topics were presented in the Ninth Conference chaired by S. Garbo.

A list of the previous nine conferences on this subject, along with their ASTM STP number and publication year, are shown below.

Previous ASTM Conferences on Composite Materials: Testing and Design

First Conference	ASTM STP 460 (1969)
Second Conference	ASTM STP 497 (1972)
Third Conference	ASTM STP 546 (1974)
Fourth Conference	ASTM STP 617 (1976)
Fifth Conference	ASTM STP 674 (1979)
Sixth Conference	ASTM STP 789 (1981)
Seventh Conference	ASTM STP 893 (1984)
Eighth Conference	ASTM STP 972 (1986)
Ninth Conference	ASTM STP 1059 (1988)

In the early conferences, all test methods for composites were in a state of flux and development, whereas, in the later conferences, including this one (tenth), the papers concentrated on new types of test methods such as Mode I interlaminar fracture toughness and the continuing problems of compression testing. In addition, testing was done using standardized test methods to obtain mechanical properties for evaluation and characterization purposes. Fracture type, failure mode identification, ultimate strength measurement, and ply/laminate failure criterion development were continued throughout all the conferences, but process control and nondestructive evaluation as part of the bigger quality assurance picture have been covered in several of the last six conferences. Analysis and design were covered in the early conferences and in several recent ones. Damage tolerance and resistance were covered in three of the last six conferences.

These conferences have given the composites technical community an ongoing detailed report of the progression of the testing and design technology of composites and its applications.

Summary

The keynote technical paper by Dr. J. Whitney on the status of selected testing technologies is followed by eight technical sessions. These sessions are summarized in the following paragraphs.

Session I on "Compression Test Methodology Analysis and Development" discusses the analysis of basic phenomena including the influence of fiber waviness on compressive behavior, structural applications of post-impact compressive fatigue of stitched composites, and the extension of basic test methodology to the development of a very high strain rate dynamic compression testing procedure. These papers illustrate that considerable progress has been made on issues related to the behavior of composites subject to compressive loading and represent worthy contributions to the advancement of the state of the art of compression testing of composites and simple structural elements.

The development and analysis of new and innovative test methodologies for composite materials are presented in *Session II* on "General Test Methodology Analysis and Development." The test methodologies discussed are thermomechanical fatigue and biaxial compression testing and the effect of tab orientation on test specimen response. These general test methodologies represent the state of the art in testing for the mechanical response of composite materials.

"Material Mechanical Properties and Failure Criteria" is the subject of the papers presented in *Session III*. These two technologies come together in developing allowables for specific structural design applications. In one paper, design properties for database use are generated on an intermediate modulus-high strain graphite/thermoplastic composite material. A discussion of the problems encountered in testing hot-wet graphite/polyimide laminates and appropriate technical solutions are presented in another paper. These presentations are followed by the presentation of a new failure criterion for composite laminate strength prediction. This criterion can be used with design data and appropriate analytical methods in the structural synthesis process.

In *Session IV*, "Advanced Materials Analysis and Test," the papers are concerned with the performance of fiber-reinforced metal, intermetallic, or ceramic matrix composites subject to static and fatigue mechanical loading at room and elevated temperatures. The mechanical response of silicon carbide fiber-reinforced titanium at elevated temperature is predicted and measured in the first paper, while a silicon carbide fiber-reinforced intermetallic matrix composite subject to elevated temperature fatigue testing is presented in the second paper. A deformation and failure prediction method using shear lag analysis is developed and experimentally

verified for brittle matrix composites in the third paper, and a macromechanics numerical analysis on a tensile loaded unidirectional composite with a central notch is presented in the last paper for metal matrix composites. These papers show that progress in testing and analysis has been made in recent years on such advanced composite materials, but they also show the need for further research.

Making a transition from the properties of materials to their structural design application in such a manner that the structure will be safe and yet exhibit the reliability, maintainability, and supportability (RM&S) that are necessary for certification is a major task. A small segment of this broad task is covered in *Session V* on “Analysis, Test, and Certification of Structure.” In the first paper, the analysis and testing of integral composite sub-elements subject to out-of-plane loadings are discussed. The design, analysis, and test of a horizontal stabilizer that is a composite-faced sandwich shell structure is presented in the second paper. In the third paper, an all-composite stub blade joint for a high-altitude, long-endurance aircraft is designed that incorporates removable composite wing structures with large spans and high aspect ratios. In the last paper, a thick quartz/polyimide V-shaped sub-element is tested in a variable load-temperature environment to measure the structural response and degradation over time. These papers illustrate some of the problems and solutions inherent in structural design and test verification of composite structure.

Session VI on “Quality Assurance and Process Control” covers the effects of processing variables, the status of automated process control, the effects of defects, and improvements in non-destructive inspection (NDI). In the first paper, co-cured sandwich panel quality variation with process changes is evaluated with destructive test procedures. The current state of the art in automated process control is presented in the second paper. This paper shows how transducers monitor the dielectric, acoustic, and fluorescent signals that are interfaced with temperature and pressure profiles during cure to determine their effect on part quality. In the third paper on the analysis of defects in composite materials, the interaction of design/analysis tools with ultrasonic nondestructive engineering data are combined to predict the performance of actual parts. Image enhancement of the ultrasonic inspection techniques allows the analyst to improve dramatically the detectability of defects. Quality assurance and process control are at the heart of acceptable quality parts and test data.

“Interlaminar Fracture Analysis and Test” are the subject of *Session VII*. Edge effects and interlaminar shear and tensile stresses and deformations are studied along with interlaminar fracture toughness and the use of adhesive film interlayers to suppress delamination in this session. Delamination from matrix cracks in multidirectional laminates and the prediction of delamination in curved unidirectional laminates are covered in the first two papers, respectively. Subsequent papers discuss fracture toughness of composite materials and the efficient use of film adhesives to suppress delamination initiation in multidirectional laminates. The technology to calculate and measure interlaminar stresses and fracture toughness properties, as well as the ability to design delamination suppression into parts, is what the composite structures design community wants and needs.

Session VIII on “Damage, Flaws, and Repairs” describes the issues that anyone designing, building, and testing composite structures must address. In the first paper, the damage tolerance of three-dimensional braided carbon/thermoplastic composites is presented. Fracture of notched laminates under loading along with its related damage tolerance are discussed in the second paper. Damage evaluation in composites using thermographic stress analysis is presented in the third paper, and the subject of composite repairs is discussed in the last paper. Increasing damage tolerance, detecting and monitoring damage, and the repair of damage are critical issues in the design application of composite materials to structures.

The Keynote technical paper plus 28 additional technical papers presented in the eight technical sessions describe the current state of the art in advanced composite materials research and test method development as well as design engineering application of composite materials

to structure. Over the last 21 years, the 10 ASTM conferences have charted the technical progress made in these areas and the resulting ASTM STP publications have become industry standard reference documents.

Glenn C. Grimes

Lockheed Advanced Development Company
(Skunk Works)
Sunland, CA; symposium chairman and editor

Keynote Address

KEYNOTE ADDRESS

James M. Whitney ¹

Reflections on the Development of Test Methods for Advanced Composites

REFERENCE: Whitney, J. M., "Reflections on the Development of Test Methods for Advanced Composites," *Composite Materials: Testing and Design (Tenth Volume)*, ASTM STP 1120, Glenn C. Grimes, Ed., American Society for Testing and Materials, Philadelphia, 1992, pp. 7–16.

ABSTRACT: The importance of analyzing failure modes in conjunction with the development of test methods is discussed. Illustrations are chosen from four classic areas of interest: flex testing of unidirectional composites exposed to moisture, the short beam shear test, the edge delamination test, and two configurations of the Illinois Institute of Technology Research Institute (IITRI) compression test method. In each of these tests, the failure mode is an important consideration in accepting and interpreting the resulting data.

KEYWORDS: failure modes, test methods, composite materials, short beam shear, edge delamination test, compression testing, compressive failure

The need to evaluate composite material test methods has increased in recent years as a result of the complex nature of data required for design considerations. In the early development of test methods for advanced composites, there was a tendency to use techniques originally developed for metals technology. However, the heterogeneous, anisotropic nature of fiber-reinforced composites required that any test method borrowed from metallic technology be carefully scrutinized before being applied directly to the characterization of composite materials. A number of papers in the technical literature addressed test methods from a theoretical standpoint.

Analysis has played an important role in the development of test methods. The work by Pagano and Halpin [1] on the end constraint problem associated with off-axis tension testing of unidirectional composites was instrumental in restricting the ASTM Test for Tensile Properties of Fiber-Resin Composites (D 3039) to orthotropic materials. The analysis by Rosen [2] on the tension testing of $[\pm 45^\circ]_s$ led to the development of ASTM Practice for In-plane Shear Stress-Strain Response of Unidirectional Reinforced Plastics (D 3518). Analytical work performed in conjunction with laminated beams [3] led to the revision of ASTM Tests for Flexural Properties of Unreinforced and Reinforced Plastics and Electrical Insulating Materials (D 790) to accommodate advanced composite materials. Other examples can be cited in which analysis played an important role in the development or revision or both of ASTM standards for composite materials.

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Although the importance of analysis in the development of experimental techniques is well recognized, attention in the present paper is concentrated in the area of failure modes associated with certain test methods. In particular, the failure mode assumed in conjunction with a given data reduction scheme may not be attained in practice. Ignoring this fact can lead to erroneous data in conjunction with material properties or material behavior.

Flex Testing of Unidirectional Composites Exposed to Moisture

Flexure tests (ASTM Tests D 790) are often used in conjunction with 0° unidirectional composites. Although the test standard refers to the calculation of “flexural strength,” the test results are often compared to tensile strength data. Thus, the failure process is assumed to be dominated by the tensile strength of the fiber. For unidirectional graphite/epoxy composites exposed to moisture, resin softening occurs at temperatures considerably below the dry glass transition temperature of approximately 175°C. Because 0° unidirectional tension is assumed to be a fiber-dominated property, moisture should have little effect on tensile strength. Although 0° graphite/epoxy composites showed little effect of moisture on tensile strength, flexural strength was dramatically reduced at moderately elevated temperatures because of the presence of moisture.

This behavior was not understood because of the assumed similarity between flexure strength and tensile strength of 0° unidirectional composites. It was shown, however, that the difference in strength was due to a change in failure mode from fiber dominated to matrix dominated (interlaminar fracture) [4]. This change in failure mode is illustrated in Fig. 1 for four-point loading of a graphite/epoxy unidirectional composite. The numbers in parentheses represent the number of samples used in obtaining each data point.

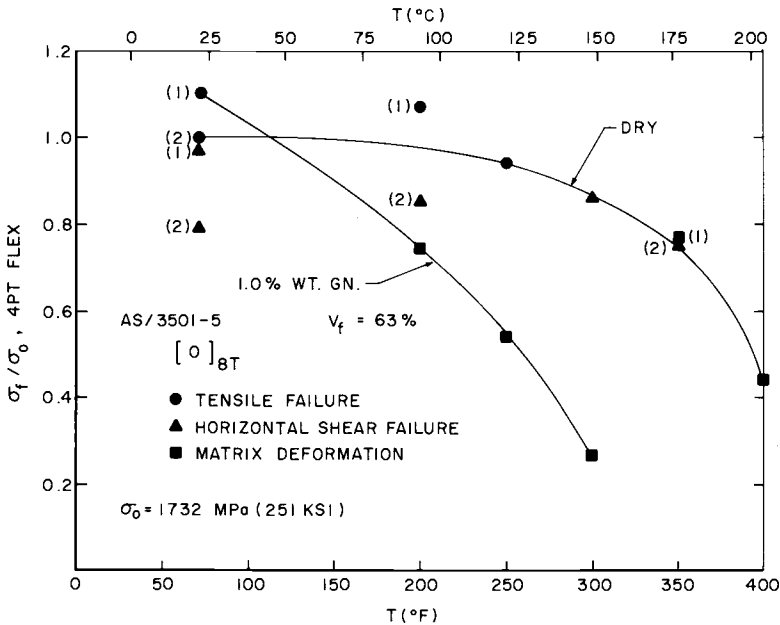


FIG. 1—Flexure data, four-point loading of AS/3501-5 specimens [4]. Numbers in parentheses indicate sample size used in obtaining the corresponding data point.

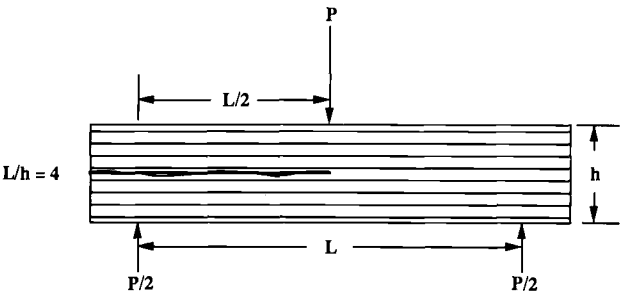


FIG. 2—Short beam shear specimen.

Short Beam Shear Tests

A classic example of the importance of failure mode can be observed in conjunction with the ASTM Test for Apparent Interlaminar Shear Strength of Parallel Fiber Composites by Short Beam Method (D 2344). As illustrated in Fig. 2, this test method involves loading a beam under three-point bending with the span-to-depth-ratio, L/h , chosen such that an interlaminar failure is induced along the centerline rather than a tensile failure in the bottom surface of the beam. ASTM Test D 2344 used in conjunction with thin 0° unidirectional beams, which is common with graphite/epoxy, usually does not yield interlaminar failure. Such data are often reported in the literature without mentioning the failure mode, leaving the reader to believe that the desired failure mode was attained. The photomicrograph in Fig. 3 [5] shows a compressive buckling failure adjacent to the load nose for a 16-ply graphite/epoxy unidirectional short beam shear specimen.

An interlaminar failure can be produced by a thick short beam shear specimen, as illustrated in Fig. 4 [5]. The horizontal split, however, is accompanied by a vertical crack. This 50-ply specimen allows the distance between the load nose and supports to be increased for a 4:1 span-to-depth ratio compared to the thin 16-ply beam specimen. It should be noted that the

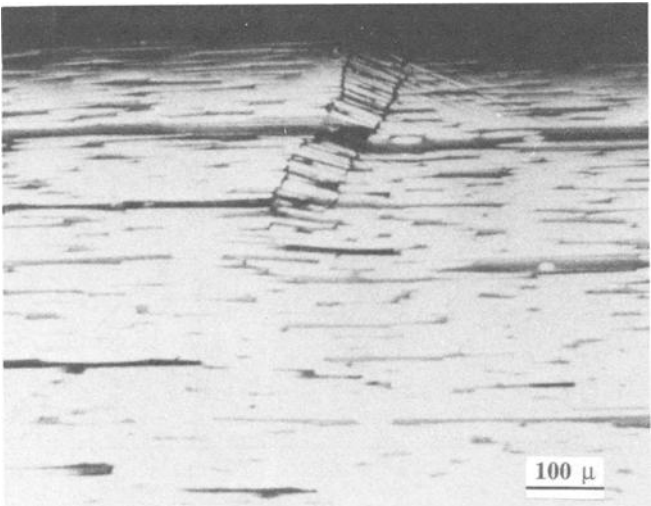


FIG. 3—Compressive failure in 16-ply AS-1/3502 graphite/epoxy unidirectional short beam shear specimen. Load applied to the right of the damage region [5].

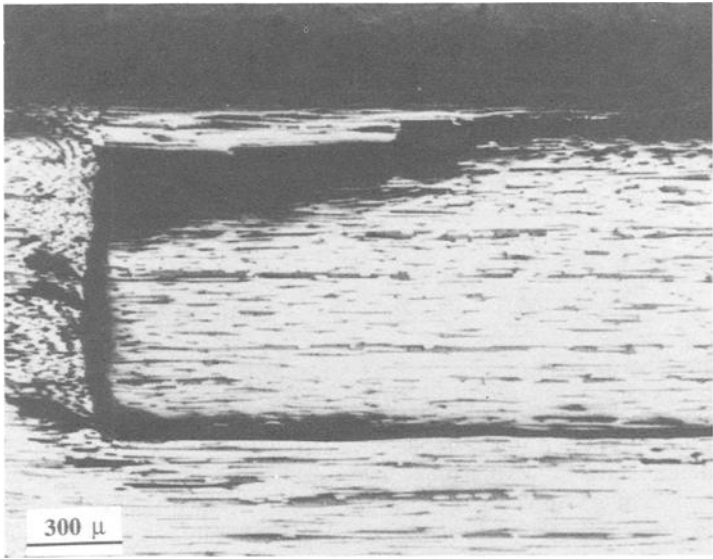


FIG. 4—Failure mode in 50-ply AS-1/3502 graphite/epoxy unidirectional short beam shear specimen. Load applied to the left of the vertical crack [5].

interlaminar failure of the 50-ply beam does not occur along the centerline but in the upper quarter of the beam.

Stress distributions obtained from classic theory of elasticity [6] are shown in Fig. 5 for the 50-ply short beam shear specimens. The normalized stress components, σ_ζ , σ_η , and $\tau_{\zeta\eta}$, represent the bending stress, through-the-thickness normal stress, and interlaminar shear stress,

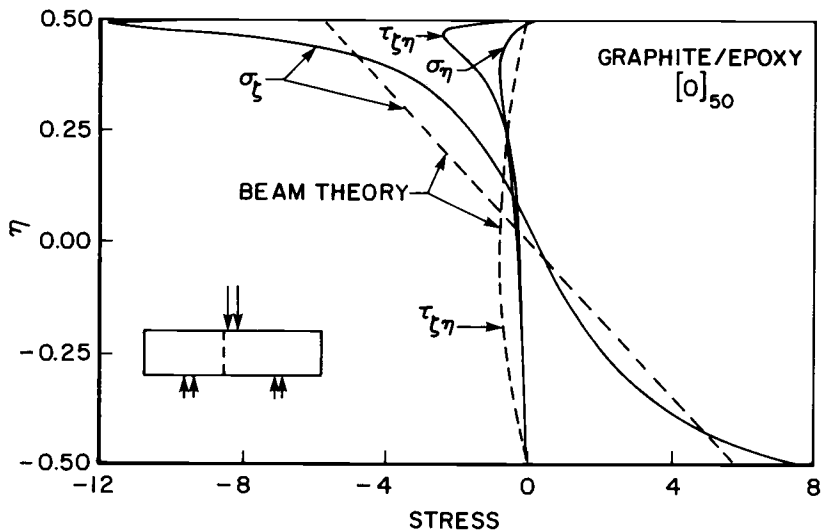


FIG. 5—Stress distribution through-the-thickness for a 50-ply short beam shear specimen, $L/h = 4$

respectively. Although not illustrated here, similar results can be obtained for the 16-ply, three-point bend specimen. The dotted lines indicate classic beam theory. It is easily seen that regions of high shear stress in the upper quarter of the beam are accompanied by biaxial compression, with the compressive stress in the fiber direction being very large. The transverse compressive stress tends to suppress shear failure. The longitudinal compressive stress is likely to be responsible for the buckling failure in the 16-ply, short beam shear specimen. The photomicrograph in Fig. 3 indicates fiber buckling at the top of the vertical crack, suggesting that these fractures were initiated by fiber buckling followed by vertical shear failure.

Note that the 16-ply specimen and the 50-ply specimen produce different values of the apparent interlaminar shear strength [5]. The data reduction equation is given by the classic short beam shear relationship.

$$\tau = \frac{3P_m}{4bh} \quad (1)$$

where

- τ = the apparent interlaminar shear strength,
- P_m = the maximum load,
- b = the specimen width, and
- h = the beam thickness.

Data reported in Ref 5 in conjunction with AS-4/3501-6 graphite/epoxy specimens yielded values of 102 and 96 MPa (14.8 and 13.9 ksi) for the apparent interlaminar shear strength of the 16- and 50-ply configurations, respectively.

Edge Delamination Test

A strain energy release rate calculation was developed by O'Brien [7] in conjunction with an edge delamination test. This test was originally thought of as a Mode I (opening mode) interlaminar fracture test. However, because the desired mode of failure was not obtained, the test became a mixed-mode interlaminar fracture test requiring finite element analysis to separate Mode I and Mode II components.

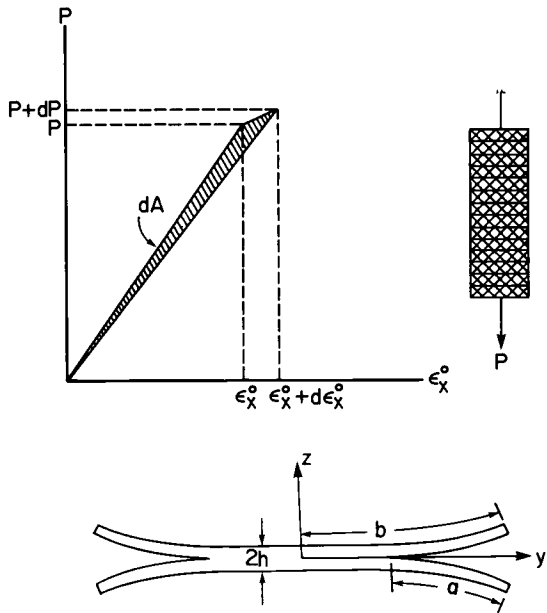
The edge delamination test uses a multidirectional laminate under tensile load with a ply-stacking sequence designed to produce free-edge delamination dominated by interlaminar tension. In his original work, O'Brien used a $[\pm 30^\circ/90^\circ/90^\circ]_s$ laminate. As shown in Fig. 6, an abrupt change in the stress-strain curve can be denoted at the onset of crack propagation. If we denote the critical value of the midplane strain ϵ_x^0 by ϵ_c , then the following strain energy release rate relationship can be derived from classic laminated plate theory

$$G_C = h\epsilon_c^2(E_x - E^*) \quad (2)$$

where

- G_C = the total value of the critical strain energy release rate,
- h = the laminate half thickness,
- E_x = the specimen modulus without a delamination, and
- E^* = the modulus of each delaminated half after the crack has propagated completely across the width of the specimen.

Although designed to measure Mode I delamination, the edge crack did not remain in the laminate midplane, but oscillated between the midplane and the $90^\circ/-30^\circ$ interfaces, pro-



$$G_{Ic} = h \epsilon_c^2 (E_1 - E_1^*)$$

FIG. 6—Free-edge delamination test.

ducing a mixed-mode fracture rather than a pure Mode I delamination. Thus, the anticipated mode of failure was not obtained. The original model was revised to reflect the double crack along each edge as observed experimentally (Fig. 7). This required that E^* be determined from the average value of the moduli determined from each delaminated section after the crack has completely propagated across the laminate width. Thus,

$$E^* = \frac{8E_x[\pm 30^\circ] + 3E_x[90^\circ]}{11} \tag{3}$$

This test method has been used with other laminate configurations. In each case, a mixed-mode fracture was produced.

Compression Tests

Compression testing has produced considerable controversy with a number of different experimental methods being proposed. Each method, however, produces different values of compression. Again, the failure mode is the key, because each test method tends to produce different failure modes. Two specimen configurations of the Illinois Institute of Technology Research Institute (IITRI) compression test are considered in which entirely different apparent compression strengths and failure modes are produced.

The IITRI test fixture for compressive loading is shown in Fig. 8 (ASTM Test for Compressive Properties of Unidirectional or Crossply Fiber-Resin Composites [D 3410]). This specimen is unsupported, with a gage section that is nominally 12.7 by 12.7 mm (0.5 by 0.5 in.). Typical data obtained in conjunction with 0° unidirectional composites can be found in Ref

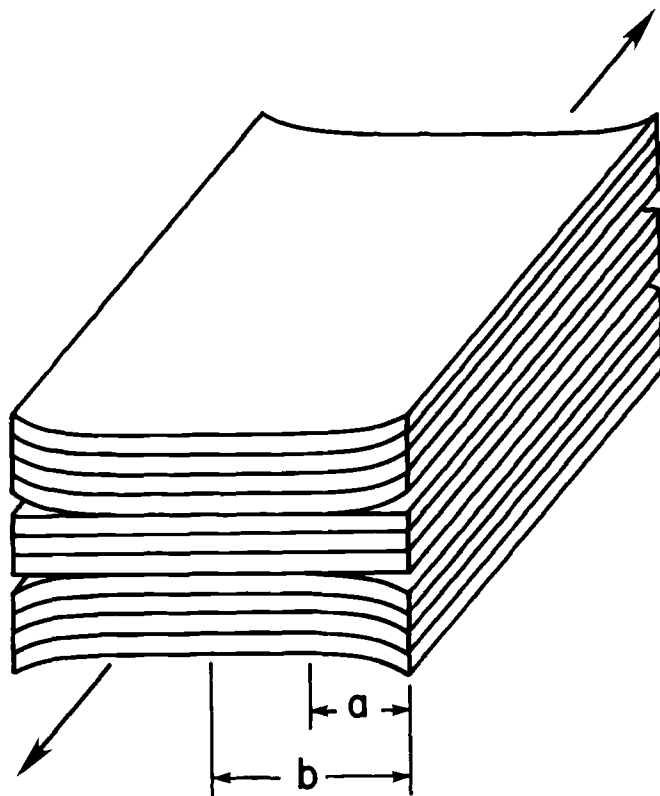


FIG. 7—Double-edge crack delamination model [7].

8. The ultimate compressive strength produced by AS-4/3501-6 graphite/epoxy material in a 16-ply configuration was 1290 MPa (187 ksi). However, back-to-back strain gages showed buckling occurred at approximately 690 MPa (100 ksi). The failed specimen is shown in Fig. 9. Considerable longitudinal splitting followed by fiber buckling was observed. A stress analysis in conjunction with the IITRI specimen performed by Tan [9] shows high stress concentrations near the end tab along with high interlaminar shear stresses under the end tab. The interlaminar shear stresses may induce delamination and longitudinal splitting.

A mini-sandwich beam specimen is shown in Fig. 10 [10]. This specimen consists of composite face sheets co-cured with a core material constructed of the matrix material contained in the face sheets. This specimen is also tested in an IITRI configuration. A typical failure is shown in Fig. 11 in conjunction with two-ply face sheets constructed of 0° unidirectional AS-4/3501-6 graphite/epoxy composites. This specimen produced a compression strength of 2070 MPa (300 ksi), almost double the value produced by the conventional IITRI specimen previously cited. The failure mode of the sandwich specimen is quite different, however. Very little longitudinal splitting is observed in Fig. 11, with fiber breakage being the primary failure mode.

Summary

Although much emphasis has been placed on the analysis of test methods and test specimens, resulting failure modes must be given serious consideration. In some cases, such as the

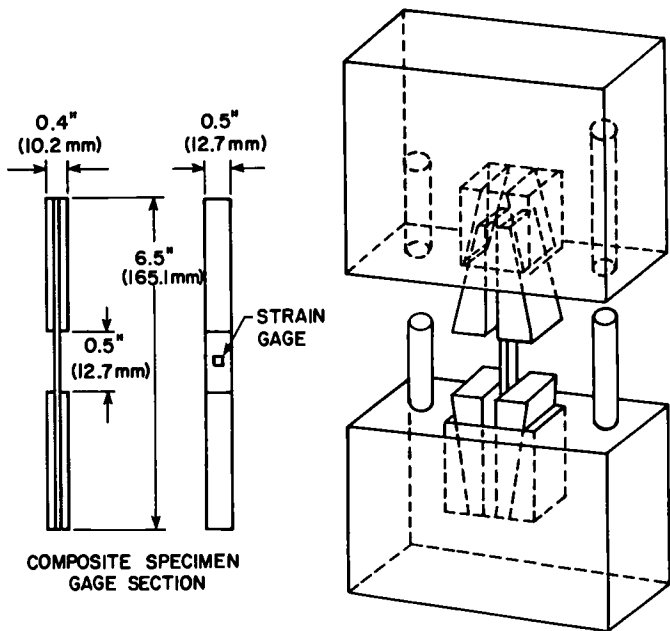


FIG. 8—The IITRI compression test specimen and fixture.

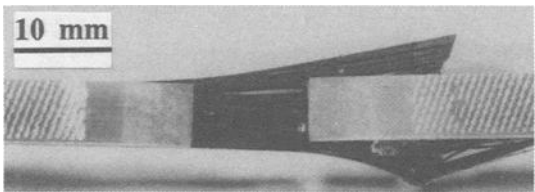


FIG. 9—Failure mode produced by AS-4/3501-6 0° unidirectional composite tested in IITRI compression.

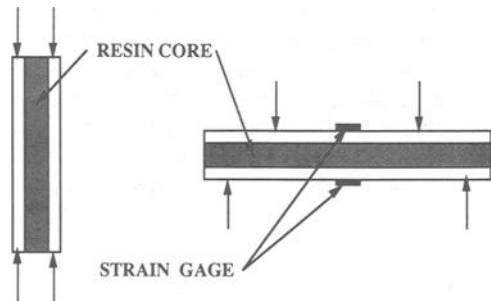


FIG. 10—Mini-sandwich beam specimen.

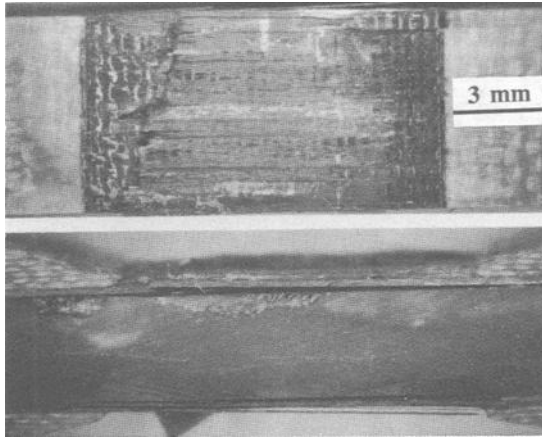


FIG. 11—Failure mode produced by IITRI compression in conjunction with mini-sandwich specimen.

short beam shear test, the desired failure mode is not produced. When data are reported, the failure mode is often ignored.

For cases such as compressive loading, analysis of the failure mode is critical. In the two examples discussed in this paper, quite different numbers and failure modes were produced by different configurations of the IITRI specimen. In analyzing failure modes, one must consider how relevant the test geometry and load introduction are to the actual application for which the data are being generated. For example, the 16-ply 0° unidirectional specimen configuration used in conjunction with the IITRI compression test promotes interlaminar fracture and longitudinal splitting. Such failure modes may not represent the behavior of the material in the application for which the data are being generated. This is particularly true because one does not expect to encounter 16 consecutive plies of 0° unidirectional material in an engineering laminate.

Examples presented in this paper clearly indicate that failure mode analysis is of primary importance when evaluating any test method. This takes on increasing importance as more complex test methods are developed.

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Session I

Compression Test Methodology Analysis and Development

Paul A. Lagace, Chairman

Overview

Considerable progress continues to be made on issues related to the behavior of composites subjected to compression—the most complicated of the loadings to understand because of the large number of possible responses and their interactions. The three papers in this section reflect the areas in which the composites community is conducting research related to compressive behavior: analysis of the basic phenomena, extensions of basic test methodology, and applications to structural concerns.

In the paper on “The Influence of Fiber Waviness on the Compressive Behavior of Unidirectional Continuous Fiber Composites,” Highsmith, Davis, and Helms search for an explanation for the sometimes “disappointing” behavior of composites in compression. They look at phenomena associated with fiber waviness, matrix nonlinearity, and free surface effects. A micromechanical model is developed to consider sinusoidal waviness of the fibers, and an experimental technique is developed to measure fiber waviness. Their results indicate that this can be an important effect, particularly when consideration is given to matrix nonlinearity. Further work such as this is necessary for the composites community to attain a working understanding of the physical phenomena at work in compressive response.

Leaving the micromechanical view and taking a global view of a composite structure, Portanova, Poe, and Whitcomb address the “Open Hole and Postimpact Compressive Fatigue of Stitched and Unstitched Carbon-Epoxy Composites.” This issue of impact and subsequent response is of particular importance in composite structures as the propagation of delamination caused by impact can severely reduce compressive load-carrying capacity. The work by these three authors shows that stitching can increase the postimpact response by keeping the delamination damage localized to the initial damage region. The authors also show that the stitching does not cause a discernible decrease in unimpacted fatigue strength, a concern caused by the “holes/inclusions” created in the stitching process. Techniques such as stitching may hold the key to increasing design allowables in composites subjected to compression.

The final paper deals with “A Method for Evaluating the High Strain Rate Compressive Properties of Composite Materials.” While many authors have dealt with quasistatic conditions, Montiel and Williams have devised a method to test composites under compression at strain rates up to 8 s^{-1} . The properties at these rates are important for a number of considerations, including certain loading conditions and impact events. The authors show that the strength of the material can be reduced by a substantial factor at these high strain rate loadings. These factors need to be taken into account in design considerations where such strain rates may be encountered.

The three papers in this session are all worth reading as they help to advance our state of knowledge concerning the response of composite materials and their structures to compressive loading.

Alton L. Highsmith,¹ John J. Davis,² and Kayleen L. E. Helms³

The Influence of Fiber Waviness on the Compressive Behavior of Unidirectional Continuous Fiber Composites

REFERENCE: Highsmith, A. L., Davis, J. J., and Helms, K. L. E., "The Influence of Fiber Waviness on the Compressive Behavior of Unidirectional Continuous Fiber Composites," *Composite Materials: Testing and Design (Tenth Volume)*, ASTM STP 1120, Glenn C. Grimes, Ed., American Society for Testing and Materials, Philadelphia, 1992, pp. 20–36.

ABSTRACT: The influence of fiber waviness and matrix nonlinearity on the compressive behavior of continuous fiber composites is studied. A micromechanics model based on the kinematics of the fibers has been developed to predict the behavior of unidirectional composites with initially wavy fibers under compressive loads. The initial waviness has been idealized as sinusoidal. Nonlinear shear behavior of the matrix has been included. Because the shear strain in the matrix is a function of position, the deformed shape of the fibers differs from the initial shape. Therefore, the deformed fiber shape is represented by a sine series.

Fiber waviness and matrix shear properties have been determined experimentally for T650-42/Radel C under several conditions. These experimental data have been used as input into the analytical model. Predicted compressive response is in reasonable agreement with experimental data.

KEYWORDS: composites, compression, fiber waviness, nonlinear shear behavior, micromechanics

Nomenclature

L	Wavelength of wavy fiber
A_0	Initial amplitude of waviness
A_1	First-order sine series coefficient
A_3	Third-order sine series coefficient
A_5	Fifth-order sine series coefficient
A_f	Cross-sectional area of the fiber
b_1	First-order matrix curve-fitting coefficient
b_2	Second-order matrix curve-fitting coefficient
b_3	Third-order matrix curve-fitting coefficient
b_4	Fourth-order matrix curve-fitting coefficient
b_5	Fifth-order matrix curve-fitting coefficient
b_6	Sixth-order matrix curve-fitting coefficient
f	Final fiber shape (function of x)

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f_0	Initial fiber shape (function of x)
E_f	Fiber modulus
ϵ	Net strain on the composite
ϵ_{avg}	Average pseudostrain
ϵ_{xx}	Actual axial strain of the centroidal axis in fiber
G_m	Matrix shear modulus (constant)
γ_m	Shear strain in matrix
I_{zz}	Fiber moment of inertia
L	Half-wavelength
P	Total potential energy
T_x	External traction in the x -direction on the composite
t	Shear stress in matrix
t_f	Half-fiber thickness
t_m	Matrix thickness
u	Fiber displacement in the x -direction
u_{fu}	Upper surface fiber displacement in the x -direction
u_{fl}	Lower surface fiber displacement in the x -direction
u_0	Centroidal fiber displacement in the x -direction
U	Total strain energy
U_f	Fiber strain energy
U_m	Matrix strain energy
u_m	Matrix displacement in the x -direction
v	Fiber displacement in the y -direction
v_0	Centroidal fiber displacement in the y -direction
v_m	Matrix displacement in the y -direction
V	External potential energy
V_f	Fiber volume fraction
W	External work done by traction T_x

Note: An i subscript on any of the above variables indicates the i th fiber or matrix region in the series model.

Introduction

Considerable effort has been expended to develop polymer matrix composites with increased toughness. Damage resistance and damage tolerance have, to a large degree, driven the development of new composite systems. Typically, an increase in composite toughness is achieved by using tougher matrix materials, for example, thermoplastics. However, this increase in toughness has not come without price. These tough resin systems that have high strains to failure tend to exhibit stress-strain behavior that is decidedly more nonlinear than that of their brittle predecessors. This often results in compressive behavior that is significantly compromised in comparison to brittle matrix (for example, epoxy) composites.

The compressive behavior of continuous fiber composites is influenced by several factors. However, two factors are believed to dominate the compressive response of unidirectional, continuous fiber-reinforced composites. One of these, nonlinear stress-strain behavior of the matrix, has already been mentioned. The second is fiber waviness, or curvature in the reinforcement. These two factors are, in fact, synergistic. The effect of these two factors on composite compressive behavior when they act in concert is significantly greater than the effects they have on composite compressive behavior individually.

The objective of this study is to develop a micromechanics model of the compressive behavior of continuous fiber-reinforced composites with nonlinear matrix behavior and fiber waviness.

response of a unidirectional composite with wavy fibers and nonlinear shear behavior of the matrix. The mathematical model for predicting the behavior of such a composite is presented herein. The fibers are treated as Euler-Bernoulli beams given transverse support by the matrix. The nonlinear shear-stress/shear-strain behavior of the matrix can be included. The model is implemented by assuming a functional form for the transverse displacement of the initially curved fibers. What sets the present model apart from other micromechanics models that account for nonlinear matrix behavior is that the functional form for transverse fiber displacement can differ from functional form used to describe the initial fiber shape. Matrix displacements are determined in terms of the fiber displacements. The principle of virtual work is used to solve for the fiber displacements.

The inputs to the model, fiber waviness and shear behavior of the matrix, have been determined experimentally. Both fiber waviness and matrix shear behavior have been measured for unidirectional T650-42/Radel C, a graphite-reinforced thermoplastic. Predicted compressive response is in reasonable agreement with experimentally observed response.

Previous Research

Numerous researchers have studied the compressive behavior of continuous fiber composites. Rosen [1] was one of the first to study the microbuckling of the fiber reinforcement. He considered unidirectionally reinforced composites with a compressive load parallel to the fiber reinforcement. Rosen also noted the importance of inelastic matrix behavior. Hayashi [2] also developed a theory based on shear instability of the matrix material, which agreed reasonably with experimental data. Greszczuk [3,4] studied a number of composite systems, and his experimental results were in good qualitative agreement with such compressive strength predictions based on fiber microbuckling.

There have also been several investigations of initially curved fibers and plies. Suarez et al. [5] examined several types of local failures of composites under compressive loads and determined that initial laminate waviness has a considerable effect on the predicted compressive strength of composite materials. Hanasaki and Hasegawa [6] studied the compressive strength of composite materials, with the assumption that the fibers are not initially straight. They further assumed that the initial curvature or waviness was parallel from fiber to fiber (that is, in-phase waviness). Using two-dimensional deformations, they developed an analytical model to compute the compressive strength of composites. The model can also consider matrix nonlinearities using secant moduli from the shear-stress/shear-strain response.

Davis [7] also studied the effects of fiber waviness on the compressive behavior of composite materials. He based his model on a repeating element of fiber surrounded by two half layers of matrix. Each layer has some initial deflection (that is, fiber waviness) represented by a sine wave. His model accounted for nonlinear shear-stress/strain behavior in the individual layers. Davis measured the waviness of several boron fibers in an epoxy matrix by cutting a specimen into several sections perpendicular to the length dimension. He measured the coordinates of five of the boron fibers at each section to approximate the shapes of the fibers through the length of the specimen. Davis also performed simultaneous compression/torsion tests on the boron epoxy specimens. The experiments showed that the shear modulus of a composite material decreased significantly as the compressive stress increased. As the shear modulus approaches zero, the composite appeared to fail due to a shear instability. His experimental results were bounded by his theoretical model for both shear modulus and compressive strength.

Lee and Harris [8] developed mathematical models for both in-phase and out-of-phase fiber waviness. The model is based on a single corrugated beam on an elastic foundation. The deformation of the surrounding material is modeled as a series of springs. Linear springs are used

for the tension-compression behavior; torsional springs are used to model the shear deformation. Euler-Bernoulli beam theory is used because the amplitude of the beam is considered to be small relative to the pitch. They used the principle of minimum potential energy to predict the final shape of the beam under a tensile load and also the strains involved in the deformation. Shuart [9,10] developed an analysis of the compressive response of multidirectional laminates that have plies that contain fiber waviness. Both in-plane and out-of-plane waviness are included. Again, material behavior is linear elastic.

Wang [11] studied the microbuckling of composites under compressive loads. His work centered on composites with some initial deflection in the fiber reinforcement, and he developed a model to predict the compressive strength of these materials. His model was also based on the idea that the in-plane shear-stress/strain relation of the composite material is nonlinear. Wang proposed that failure occurred when the load caused an unstable increase in the fiber deflection, which corresponds to a large increase in the in-plane shear stress. His model considered the initial fiber curvature to be a statistical material property that must be determined for use in the model. The shear-stress/strain behavior of the material must also be known for the model. Wang determined the initial fiber waviness by comparing analytical predictions of compressive strength to those obtained experimentally at room temperature. He used this initial fiber waviness parameter to predict the compressive strength at various other temperatures. He assumed that an increase in temperature would only change the shear behavior of the composite and not the initial waviness. He found the waviness (amplitude/wavelength) value to be 0.0108, and he reported good agreement between his model and experiment for the other temperatures.

Hahn and Williams [12] also developed a microbuckling model that includes matrix nonlinearity. Again, the fibers are treated as linear elastic beams. However, this analysis assumes a uniform shear strain in the matrix material.

Of the studies described above, only Davis [7] measured fiber waviness. However, his sectioning process for determining fiber waviness is not well suited to smaller diameter fibers, such as graphite fibers, because it is quite difficult to locate the same fiber in successive sections.

Observations

Before discussing the analytical model, it is appropriate to make a few observations about fiber waviness and to make note of the approach to measuring fiber waviness used in the present study. Compression tests were performed at General Dynamics/Fort Worth Division. Unidirectional T650-42/Radel C laminates 2 by 2 ft (0.61 by 0.61 m) and six plies thick were fabricated. Each of these test skins was incorporated into a sandwich panel along with an aluminum honeycomb core (23 lb/ft³ [369 kg/m³], 0.125-in. [3.18-mm] hex cell, 1.5 in. [38.10 mm] deep) and a 15-ply unidirectional graphite/epoxy reaction skin, which were bonded together using an adhesive. Axial sandwich beams 1 in. (25.4 mm) wide were cut from these panels.

Mechanical testing was performed in an Instron screw-driven testing machine. The sandwich beams were loaded in four-point bending, with the test skins on the compressive side of the beam. The supported span was 18 in. (0.457 m) long; the test section was 4 in. (0.102 m) long. Load cell output and strain data from two axial strain gages bonded to the test skin within the test section were recorded using a computerized data acquisition system.

A visual inspection of the test skins revealed numerous small "wrinkles" in the thermoplastic composite. These wrinkles were regions in which the fiber reinforcement was quite wavy. A schematic representation of these wrinkles within a specimen is presented in Fig. 1. For the most part, the fibers are relatively straight. However, there are distinct regions within the panel where the fibers are quite wavy. Within the wrinkles, the waviness is almost exclu-

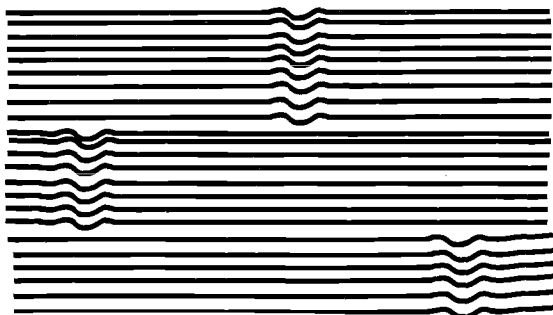


FIG. 1—Schematic representation of fiber waviness within a unidirectional panel.

sively in-phase, that is, neighboring fibers trace out nearly parallel paths. Further, the wrinkles extend through the entire thickness of the test skin. Although at first glance these wrinkles appeared to indicate out-of-plane waviness, in the specimens considered in the present study the fiber waviness occurred predominantly within the plane of the panel. An inspection of samples of prepreg tape did not reveal the existence of similar pronounced wrinkles, indicating that the wrinkles within the test skins developed or grew during processing.

In addition, the wrinkles had a decisive role in determining the location of the compressive failure. Fracture occurred at wrinkled regions. When assessing the mechanical response of such a material, note that the straight regions are relatively stiff and strong, and the wavy regions are relatively compliant and weak. Under compressive loading, composite strength is limited by the strength of the wrinkled regions.

An important question is how to quantify the waviness of a fiber. Typically, a curved fiber shape is idealized as a sinusoidal form (Fig. 2). Note that the curved shape is characterized by two length parameters—the wavelength, or pitch, L , and the amplitude A_0 . The waviness of a fiber is represented by the ratio of these parameters, A_0/L . Because the amplitude and wavelength differ by two or more orders of magnitude, the experimental determination of waviness is challenging.

The approach chosen for measuring waviness is as follows. First, a unidirectional specimen is polished, and then lightly etched in an acid solution. Next, the prepared surface is photographed using a relatively low magnification optical microscope. For this study, the magnification on a 35-mm negative was approximately $\times 12$. This negative is then used to produce a 16- by 20-in. (406- by 508-mm) print. In this enlarged view, both the wavelength and the amplitude can be measured.

Figure 3a is a photomicrograph taken from a region within a unidirectional panel of T650-42/Radel C fabricated using a graphite tool. Within the region shown, the fibers are relatively straight. However, the fiber waviness is not zero. It can be measured by tracing along a fiber and recording specific points using a digitizer. A Hewlett-Packard 7475A plotter equipped with a fiber optic sight instead of a pen is well suited to this process, because the plotter can maintain the position of the sight. The digitized points along a fiber can then be fit with a sine function to determine the amplitude and wavelength of the wavy fiber. The amplitude and wavelength differ by about three orders of magnitude in this region. At present, there are insufficient data available to give a detailed description of fiber waviness in these regions of relatively straight fibers. However, for these regions the waviness parameter is typically on the order of 0.005.

Figure 3b is a photomicrograph taken from another region within the same specimen. The curvature of the fibers is quite obvious. Note that the fiber waviness is in-phase, that is, the

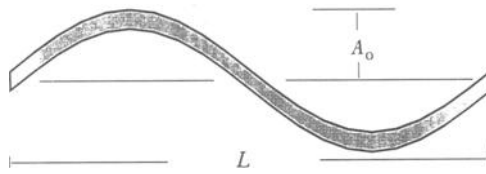


FIG. 2—Idealized representation of a wavy fiber.

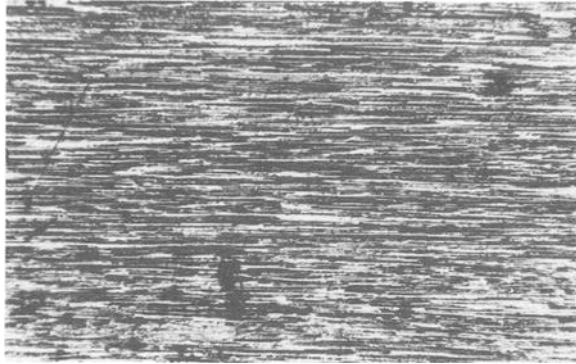


FIG. 3a—Photomicrograph of T650-42/Radel C specimen taken from a region of straight fibers in a panel fabricated against a graphite tool.

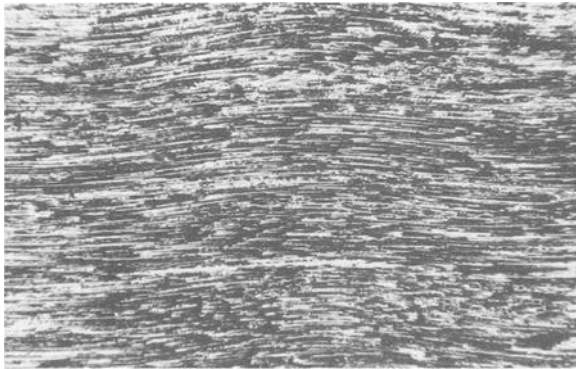


FIG. 3b—Photomicrograph of T650-42/Radel C specimen taken from a region of wavy fibers in a panel fabricated against a graphite tool.

fibers within this region follow generally the same path, which is indicative of some kind of cooperative motion by the fibers. Once again, note that this waviness is not present in the prepreg before fabrication. This waviness is almost exclusively an in-plane phenomenon. The fibers do not wander significantly through the thickness of the composite. There are regions in which the fibers terminate, which suggests that there is a small amount of out-of-plane waviness. These fiber terminations make it difficult to trace a single fiber all the way across an enlarged photograph. However, it is possible to follow a cluster of fibers across the photograph. In the complete print, peaks and valleys associated with the generally sinusoidal shape of the fibers can be identified. Further, within the print, the amplitude is of sufficient size to be measured with a ruler. For the region depicted, the wavelength was found to be 0.090 in. (2.29 mm), and the amplitude was found to be 0.002 07 in. (0.0526 mm). Thus, the waviness parameter is found to be 0.023.

A number of wrinkles in the T650-42/Radel C material fabricated against a graphite tool were studied, and all of these wrinkles were found to have similar geometric characteristics. All of the wrinkles exhibited about the same wavelength, amplitude, and waviness. This suggests that the distribution of waviness within a specimen is essentially bimodal, with one waviness mode characterizing the regions of straight fibers, and a second waviness mode characterizing the wrinkled regions. Area measurements taken from these specimens indicate that wrinkles comprise about 2% of the area, and hence 2% of the volume, of these test skins.

Another interesting observation involves the influence of the tool used during processing on fiber waviness. Figure 3c is a photomicrograph of a wavy region from a panel of T650-42/Radel C fabricated against a titanium tool. The waviness is noticeably greater than that found in the specimen fabricated against a graphite tool (Fig. 3b). For the panel fabricated against a titanium tool, the wavelength was found to be 0.0755 in. (1.918 mm), and the amplitude was found to be 0.003 69 in. (0.0937 mm), which results in a waviness of 0.0489. This waviness is more than double that found in the panel fabricated against a graphite tool. Once again, different wrinkled regions within specimens fabricated against the titanium tool had similar waviness parameters. However, a much larger fraction of the composite (approximately 10%) was composed of wrinkled regions. Regions in which the fibers were relatively straight had essentially the same waviness as corresponding regions in material fabricated against the graphite tool. The process by which fiber waviness develops is not sufficiently well understood to explain in detail this difference in fiber waviness. The most obvious difference between the tools is that there is a greater mismatch between the thermal expansion coefficient of the composite and the titanium tool.

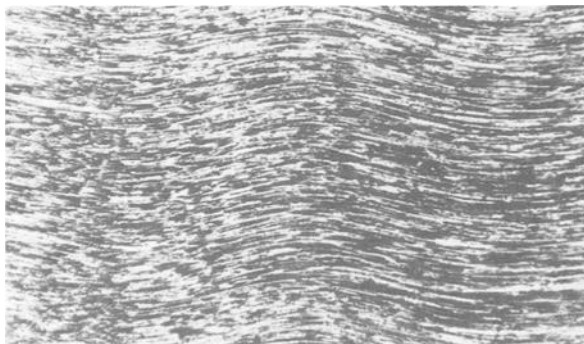


FIG. 3c—Photomicrograph of T650-42/Radel C specimen taken from a region of wavy fibers in a panel

fabricated against a titanium tool.

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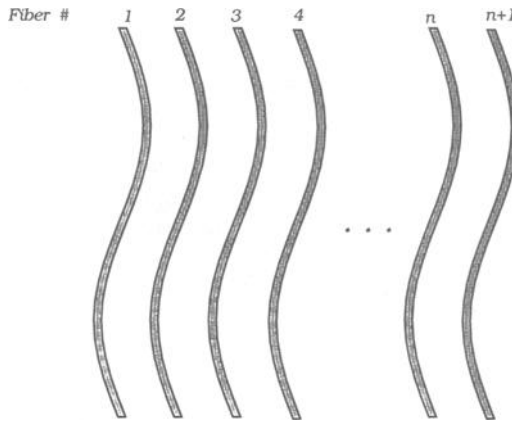


FIG. 4—Schematic representation of an array of wavy fibers.

T650-42/Radel C was the material chosen for this investigation because it seems to exhibit extreme fiber waviness. The poor compressive behavior of this system offers a critical assessment of the predictive capabilities of the analytical model.

Analysis

To predict the compressive behavior of a unidirectional composite with wavy fibers, an array of fibers with in-phase waviness has been modeled. Such an array is represented in Fig. 4. All fibers have identical initial waviness, which is given by

$$f(x) = A_0 \sin\left(\frac{2\pi x}{l}\right) \quad (1)$$

Under compressive loading, significant shear strains develop in matrix material at the nodal points. It is assumed that the composite fails when there is local failure of the matrix at these points of high shear. The failure scenario is that once there is local failure of the matrix, the fibers lose their transverse support and buckle. Failure in the regions of high waviness causes the fibers in adjacent regions of low waviness to be overloaded, and hence the composite fails.

For the present study, an infinite array of such fibers has been modeled. For an infinite array, all fibers experience the same stresses, strains, and displacements. It is only necessary to analyze a representative volume, which includes a single representative fiber. Figure 5 shows such

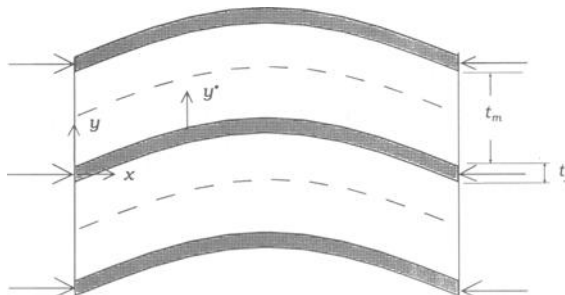


FIG. 5—A representative volume.

a representative volume bounded by the dashed lines. Note that only one half of the wavelength is included. Two transverse coordinates are indicated. Coordinate y is the local coordinate for the fiber; coordinate y^* is the local coordinate for the matrix. The symmetry of the problem suggests that the positive and negative lobes of the sinusoidally shaped fibers should deform identically, and thus only half of the wavelength is required for the analysis. The fiber thickness t_f is chosen to be the diameter of the fiber, and the matrix thickness t_m is chosen such that the fiber spacing matches the fiber spacing in the composite. Idealizing the three-dimensional arrangement of the fibers square array, the fiber volume fraction is given by

$$V_f = \left(\frac{t_f}{t_f + t_m} \right)^2$$

As will be seen in the subsequent development, the matrix shear strain, which controls predicted compressive strength, depends directly on the fiber spacing.

The analysis requires a description of the kinematics of the representative fiber. The kinematics of the matrix regions are determined in terms of the kinematics of the fiber. The principle of virtual work is used to determine explicitly the displacements in the fiber.

Fiber Kinematics

As noted earlier, the fiber has an initial shape given by

$$f_0(x) = A \sin \left(\frac{2\pi x}{L} \right)$$

Under compressive loading, the curvature of fiber causes bending moments to be developed within the fiber. Thus, the fiber will assume a new shape, given by

$$f(x) = f_0(x) + v_0(x) \tag{2}$$

where $v_0(x)$ is the transverse displacement of the centroidal axis of the fiber.

The axial displacement of the fiber is composed of three parts. First, the fiber carries an axial load, which gives rise to an axial strain, ϵ_0 . Second, because the fiber also carries a bending load, there is a corresponding bending component of axial displacement, which varies linearly through the thickness of the fiber. Finally, this bending deformation changes the end-to-end length of the fiber. This displacement component, caused by the coiling of the fiber, can be represented by a pseudostrain, given by

$$\epsilon_{ps} = \frac{1}{2} \int_0^L [(f')^2 - (f_0')^2] dx \tag{3}$$

For the fiber, the axial displacement is given by

$$u(x, y) = \epsilon_0 x - y \frac{dv_0}{dx} + \epsilon_{ps} x \tag{4}$$

Note that the pseudostrain ϵ_{ps} does not contribute directly to local strains within the fiber, which have the form

$$\epsilon(x, y) = \epsilon_0 - y \frac{d^2 v_0}{dx^2} \quad (5)$$

but does enter the analysis when determining the virtual work done by the applied traction.

Matrix Kinematics

As can be seen in Fig. 5, the matrix surrounding the fiber of interest (as indicated by the dashed lines) is equivalent to the matrix in a single cell bounded by successive fibers. For simplicity, this single matrix cell is considered. Further, the displacements of this matrix must match the displacements of the fiber at the fiber/matrix interface. The central fiber is identified as fiber i , while the top fiber is identified as fiber $i + 1$. The matrix displacements, which must match the fiber displacements at the cell boundaries, are assumed to vary linearly in the transverse coordinate, between the displacement at the top of fiber i and the displacement at the bottom of fiber $i + 1$. Thus, the matrix displacements are given by

$$\begin{aligned} u^m(x, y^*) &= u^i(x, t_f/2) + \frac{y^*}{t_m} [u^{i+1}(x, -t_f/2) - u^i(x, t_f/2)] \\ v^m(x, y^*) &= v_0^i(x) + \frac{y^*}{t_m} [v_0^{i+1}(x) - v_0^i(x)] \end{aligned} \quad (6)$$

where the superscripts i , $i + 1$, and m denote fiber i , fiber $i + 1$, and matrix, respectively.

Equation 6 can be simplified by noting that for an infinite array of fibers, fiber i and fiber $i + 1$ experience the same deformation. Hence,

$$\begin{aligned} u^m(x, y^*) &= u(x, t_f/2) + \frac{y^*}{t_m} [u(x, -t_f/2) - u(x, t_f/2)] \\ v^m(x, y^*) &= v_0(x) \end{aligned} \quad (7)$$

where u and v are the axial and transverse displacements of the fiber as specified in Eqs 4 and 2, respectively.

Virtual Work

An approximate solution to the infinite array problem is obtained by assuming a functional form for the transverse displacement of the representative fiber. Three different models have been developed, which use three different expressions for the transverse displacement. The first model uses a transverse displacement with the same functional form as the initial waviness. This form for the displacement is essentially a one-term sine series:

$$v_0(x) = A_1 \sin\left(\frac{2\pi x}{L}\right) \quad (8)$$

One difficulty with this form for the displacement is that it does not allow for changes in the shape of the fiber. Such shape changes are expected when the matrix exhibits nonlinear shear behavior, because the shear strain is not uniform throughout the matrix.

An obvious generalization of the one-term transverse displacement, which accommodates changes in the shape of the fiber, is a complete sine series:

$$\nu_0(x) = \sum_{n=1}^{\infty} A_n \sin\left(\frac{2n\pi x}{L}\right)$$

However, the symmetry of the problem requires that only the terms where n is odd be included. For the present study, both a two-term expansion:

$$\nu_0(x) = A_1 \sin\left(\frac{2\pi x}{L}\right) + A_3 \sin\left(\frac{6\pi x}{L}\right) \quad (9)$$

and a three-term expansion:

$$\nu_0(x) = A_1 \sin\left(\frac{2\pi x}{L}\right) + A_3 \sin\left(\frac{6\pi x}{L}\right) + A_5 \sin\left(\frac{10\pi x}{L}\right) \quad (10)$$

were used. When the principle of virtual work is applied, the virtual displacements have corresponding functional forms:

$$\text{One-term: } \nu_o(x) = \delta A_1 \sin\left(\frac{2\pi x}{L}\right)$$

$$\text{Two-term: } \nu_o(x) = \delta A_1 \sin\left(\frac{2\pi x}{L}\right) + \delta A_3 \sin\left(\frac{6\pi x}{L}\right) \quad (11)$$

$$\text{Three-term: } \nu_o(x) = \delta A_1 \sin\left(\frac{2\pi x}{L}\right) + \delta A_3 \sin\left(\frac{6\pi x}{L}\right) + \delta A_5 \sin\left(\frac{10\pi x}{L}\right)$$

Note that the coefficients δA_i represent the degrees of freedom in the problem.

Given the functional form for the fiber displacement, it is possible to determine functional forms for the strains in both the fiber and matrix. In a similar fashion, virtual strains can be determined. Using appropriate constitutive relationships for the fiber and matrix, the stresses in both the fiber and matrix can also be determined, and the principle of virtual work can be applied. In the expressions that follow, a unit depth has been assumed. For the fiber, which is treated as a linear elastic beam, the internal virtual work has the form [13]:

$$\delta W_f^i = - \int_0^L \int_{A_f} [E_f \epsilon(x, y) \delta \epsilon(x, y)] dA \, dx \quad (12)$$

where

E_f = the modulus of the fiber,

A_f = the cross-sectional area of the fiber, and

$\epsilon(x, y)$ is given by Eq 5.

Within the matrix, axial strains are neglected. The shear strain in the matrix is given by

$$\gamma_m(x, y) = \frac{\partial u^m}{\partial y} + \frac{\partial v^m}{\partial x}$$

and the internal virtual work in the matrix has the form

$$\delta W_I^m = - \int_{V_m} \tau^m(x, y) \delta \gamma^m(x, y) dV \quad (13)$$

where

$\tau^m(x, y)$ = the shear stress in the matrix, and
 V_m = the volume of the matrix cell.

The explicit determination of δW_I^m requires a shear-stress/shear-strain relationship for the matrix material. If the matrix material is assumed linear elastic, this relationship has the form

$$\tau_m(\gamma_m) = G\gamma_m \quad (14)$$

where G is the shear modulus of the matrix. If the matrix exhibits linear shear behavior, the one-term description of the transverse fiber displacement provides an exact solution to the problem. However, with thermoplastic matrix material, the shear behavior is decidedly nonlinear.

A method proposed by Guynn et al. [14] was used to determine the matrix shear behavior from the shear behavior of the composite. A shear-stress/shear-strain curve for the composite is determined from a $[\pm 45]_s$ laminate tension test. Then, using fiber properties, the fiber volume fraction, and micromechanics, the shear-stress/shear-strain curve for the matrix material is determined. Figure 6 shows plots of shear stress versus shear strain for the composite at two different conditions—room temperature dry and 275°F (135°C) wet. Figure 7 shows the corresponding plots of shear stress versus shear strain for the matrix material.

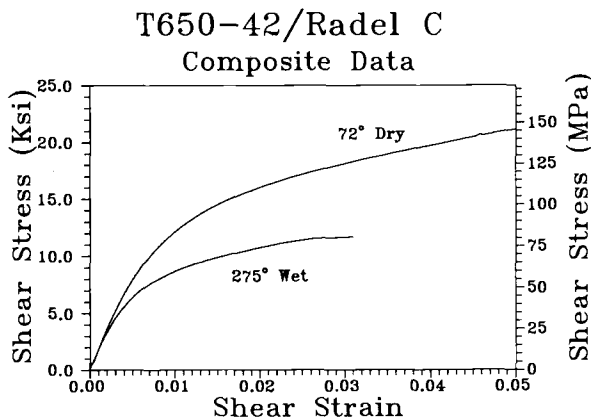


FIG. 6—Shear-stress/shear-strain curves for T650-42/Radel C at room temperature-dry and hot-wet conditions.

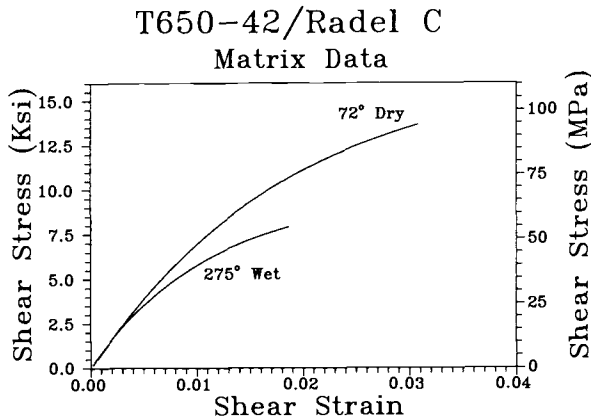


FIG. 7—Shear-stress/shear-strain curves for Radel C matrix material at room temperature-dry and hot-wet conditions.

The two curves shown in Fig. 7 were fitted with a sixth-order polynomial of the form

$$\tau(\gamma_m) = \sum_{k=1}^6 b_k \gamma_m^k \tag{15}$$

This stress strain relationship was used in Eq 13 to determine δW_I^m for the case of nonlinear shear behavior of the matrix. The coefficients for the polynomial fits of the two shear-stress/shear-strain curves in Fig. 7 are presented in Table 1. As a note, a polynomial fit is not a natural choice for representing the kind of shear-stress/shear-strain curves shown in Fig. 7. However, such a representation was chosen because of the ease of evaluating trigonometric functions raised to integer powers.

The external virtual work is given by

$$\delta W_E = \bar{\sigma}(t_f + t_m)L(\delta \epsilon_0 + \delta \epsilon_{ps}) \tag{16}$$

where $\bar{\sigma} = E_f \epsilon_0$, the net applied stress. The governing equations now take the form

$$\delta W_I^f + \delta W_I^m + \delta W_E = 0 \tag{17}$$

A numerical solution is obtained by applying a net strain ($\epsilon_0 + \epsilon_{ps}$) to the model. The resulting system of nonlinear equations can be solved using Newton's method, to determine the coefficients A_i in the assumed form for the transverse displacement, and the axial strain caused by

TABLE 1—Polynomial coefficients and ultimate shear strain data for Radel C matrix material at room temperature dry and hot-wet conditions.

Conditions	b_1 , Msi (GPa)	b_2 , Msi (GPa)	b_3 , Msi (GPa)	b_4 , Msi (GPa)	b_5 , Msi (GPa)	b_6 , Msi (GPa)	ϵ_{ult} , %
RTD	0.880 02 (6.067 7)	−21.50 (−148.24)	305.08 (2103.5)	−2 363.6 (−16 297)	9 186.1 (63 338)	−13 534 (−93 317)	3.0
275°F wet (135°C wet)	0.871 70 (6.010 4)	−38.11 (−262.77)	1001.50 (6905.3)	−15 009.0 (−103 487)	119 980 (827 262)	−403 590 (−2 782 750)	1.8

the axial load in the fiber ϵ_0 . With these parameters determined, stresses and strains in both the fiber and the matrix can be calculated.

Predicting Composite Response

As stated previously, the philosophy used for predicting the compressive behavior of the composite is that the strength of the composite is controlled by regions where the waviness is large, whereas the stiffness of the composite is controlled by some average value of waviness. To predict the strain to failure of the composite, the applied strain discussed above is incremented, and the resulting system of nonlinear equations is solved. The maximum shear strain in the matrix, which occurs where x is zero, is calculated. When this maximum shear strain exceeds the ultimate shear strain shown in Fig. 7, composite failure is predicted.

Next, the compressive stress-strain curve for the composite is predicted. Here, an average fiber waviness is used. Again, the applied strain is incremented, but the strain increments are stopped when the strain to failure determined from the analysis of the wavy region is achieved.

Results and Discussion

Predictions of compressive behavior were made using the matrix properties shown in Table 1, along with a fiber modulus E_f of 42.0 Msi (290 GPa), a fiber thickness t_f of 0.000 30 in. (7.62 μm), and a matrix thickness t_m of 0.000 13 in. (3.30 μm). Figure 8 shows the predictions of maximum shear strain in the matrix in a wavy region within a T650-42/Radel C panel fabricated against a graphite tool as a function of applied axial strain for the various models. Note that the ultimate shear strain for the matrix material is 3.0%. The model that uses linear elastic matrix shear behavior does not achieve this maximum shear strain within the plot. The response exhibits a slight nonlinearity as a result of the curvature of the fiber. For the models that use nonlinear matrix shear behavior, the response is decidedly nonlinear, and the maximum shear in the matrix is achieved within the confines of the plot. The difference between the predictions of the linear elastic matrix model and the nonlinear matrix models illustrates that matrix nonlinearity has a significant influence on the compressive behavior of a unidirectional composite. Further, while the response predicted by the one-term model differs from the response predicted by the two- and three-term models, there is virtually no difference

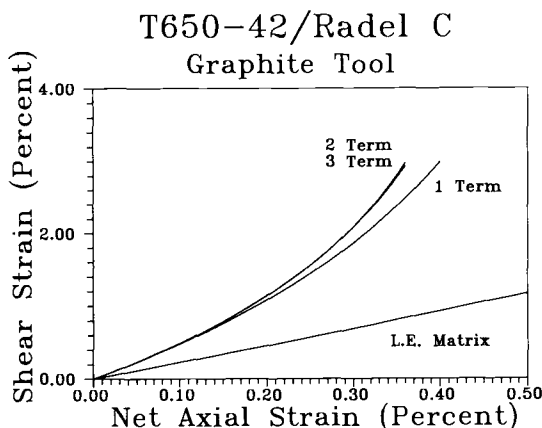


FIG. 8—Predictions of maximum shear in the matrix versus applied compressive strain for T650-42/Radel C fabric loaded against a graphite tool and loaded under room temperature-dry conditions.

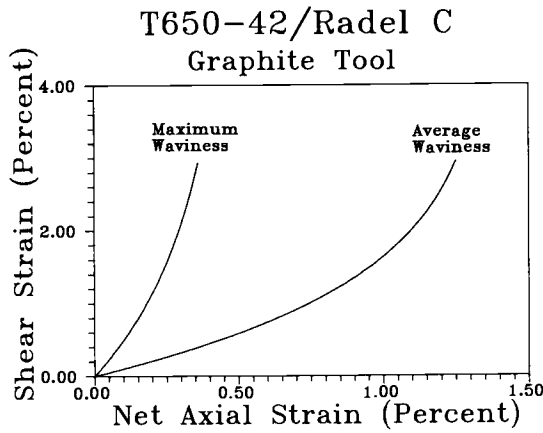


FIG. 9—Predictions of maximum shear in the matrix versus applied compressive strain for wavy material and material with average waviness.

between the response predicted by the two- and three-term models. In fact, the two- and three-term models consistently produced results that were essentially identical, which suggests that the two-term model is sufficient for predicting composite response. Composite failure is predicted when the applied compressive strain reaches 0.36%.

As discussed previously, the shape of the compressive stress-strain curve should be determined by an average value of waviness. As yet, there are not sufficient experimental data to determine an average waviness. The average waviness has been estimated using a simple “rule of mixtures” approach. The average waviness calculated using the waviness of the straight regions (waviness = 0.005), fraction of composite corresponding to straight regions (0.98), the waviness of the wrinkled regions (0.023), and the fraction of the composite that was wrinkled (0.02), was 0.0054. To illustrate the effect of fiber waviness on compressive response, Fig. 9 presents predicted maximum matrix shear strain as a function of applied axial strain for regions of high waviness and average waviness. Once again, room temperature dry matrix properties and the two-term model are used. The model predicts that if the regions of high waviness could be eliminated, a much higher applied strain would be required to cause shear failure of the matrix material. In fact, the applied compressive strain to failure for the composite is predicted to increase to about 1.25%.

Figure 10 shows both predicted and experimental compressive stress-strain data for the T650-42/Radel C material fabricated against a graphite tool, loaded under room temperature dry conditions. The experimental data were obtained from a sandwich beam four-point bend test, which was described previously. Three replicate tests were performed. The data points shown are from the test that produced the intermediate strength value. Predicted stress-strain curves for both the wavy region and for average waviness are presented. Composite failure is expected when the shear strain in the matrix in a wavy region exceeds the ultimate shear strain for the matrix. The predicted strain to failure is in excellent agreement with the experimental value. The stress-strain curve determined using the average waviness underpredicts the stiffness and strength. This is likely due to a mismatch between the fiber volume fraction used in the analysis ($V_f = 0.5$) and the actual fiber volume fraction of the composite.

Several predictions of compressive strain to failure have been made. Composite panels fabricated against a graphite tool and tested at room temperature dry conditions have been ana-

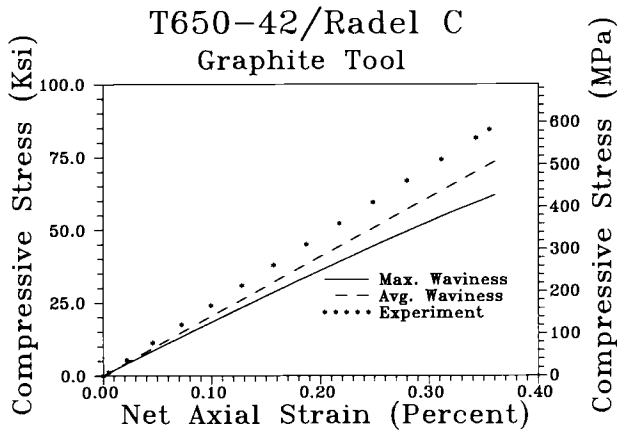


FIG. 10—Predicted and observed compressive stress-strain response of T650-42/Radel C fabricated against a graphite tool and loaded under room temperature-dry conditions.

lyzed and treated as a reference material. In addition, composite panels fabricated against a graphite tool and tested at 275°F (135°C) have been analyzed. This material should have the same distribution of waviness as the reference material. However, the matrix properties are quite different. Finally, composite panels fabricated against a titanium tool and tested at room temperature dry conditions have been analyzed. In this case, the matrix properties are the same as those for the reference material, but the fiber waviness is different. Pertinent data are shown in Table 2. Again, the experimental values of compressive strain to failure are averages from three replicate tests. Changing either the fiber waviness or the matrix properties has a significant effect on the compressive strain to failure. Also, the agreement between model predictions and experiment is quite reasonable.

Conclusions

Several theoretical models have been developed for predicting the compressive response of unidirectional composites with wavy fibers. Nonlinear shear behavior of the matrix material must be included in the analysis. Further, the two-term description of the transverse displacement of the fiber appears to be sufficient to describe the fiber kinematics.

Both fiber waviness and matrix shear-stress/shear-strain data have been determined experimentally for T650-42/Radel C, a thermoplastic matrix reinforced by continuous graphite

TABLE 2—Comparison of predicted strain to failure with experiment for different T650-42/Radel C specimens.

Tool	Conditions	L , in., (mm)	A_0 , in. (mm)	Ultimate Strain Prediction, %	Ultimate Strain Experiment, %
Graphite	RTD	0.090 (2.29)	0.002 07 (.052 6)	0.36	0.356
Graphite	275°F wet (135°C wet)	0.090 (2.29)	0.002 07 (.052 6)	0.21	0.234
Titanium	RTD	0.0755 (1.92)	0.003 69 (.093 7)	0.27	0.299

fibers, under several conditions. These experimental data have been used to predict the compressive behavior of the unidirectional composite under several conditions. Predictions of compressive strain to failure are in reasonable agreement with the experiment. The two-term model with nonlinear matrix behavior appears to be useful for assessing the effect of altering the matrix material or the fiber waviness on compressive response.

Because this relatively simple model seems capable of predicting compressive behavior, future work should focus on adjusting material parameters that influence compressive behavior. Specifically, the cause of fiber waviness merits additional study. As yet, the process by which fiber waviness develops during forming operations is not well understood. Of particular interest is how constraint, offered either by the tooling or by adjacent plies in a laminate, might be used to limit fiber waviness, and thus provide enhanced compressive behavior.

Acknowledgments

The authors gratefully acknowledge support of the modeling effort from NASA Langley Research Center under NASA Grant NAG-1-711, and support of the experimental effort from General Dynamics/Fort Worth Division.

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Open Hole and Postimpact Compressive Fatigue of Stitched and Unstitched Carbon-Epoxy Composites

REFERENCE: Portanova, M. A., Poe, C. C., and Whitcomb, J. D., "Open Hole and Postimpact Compressive Fatigue of Stitched and Unstitched Carbon-Epoxy Composites," *Composite Materials: Testing and Design (Tenth Volume)*, ASTM STP 1120, Glenn C. Grimes, Ed., American Society for Testing and Materials, Philadelphia, 1992, pp. 37-53.

ABSTRACT: This investigation evaluates the fatigue performance of a stitched uniweave fabric composite and a toughened tape composite. It also addresses the effects of stitching on compression fatigue life. Postimpact compressive fatigue and open hole fatigue tests were run on an AS4/3501-6 uniweave with stitching and a toughened IM7/8551-7 tape without stitching. Stitching was found to increase the thickness and, consequently, the weight of the composite material. The two materials were compared on an equal carbon content basis as well as on an equal weight basis. The excess thickness in the stitched uniweave composite was responsible for the lower fatigue life, on an equal carbon basis, compared to the toughened resin tape composite. Comparison of fatigue lives on an equal carbon fiber areal weight basis indicated that puncture or crimp damage from stitching has very little effect on compression failure. Postimpact fatigue tests showed that, although the damage area in the stitched uniweave composite was twice that of the toughened tape composite, the fatigue lives of the stitched composite were significantly longer than those of the toughened composite. Thus, it appears that the increase in thickness from stitching is much more of a penalty than crimped fibers or puncture damage from stitching.

KEYWORDS: delamination, open hole fatigue, postimpact fatigue, stitched composites, through-the-thickness stitching, textile composites, low-velocity impact

Compressive strength of laminated composites can be significantly reduced by delamination. Delamination can develop at a free edge, particularly at a hole, where interlaminar stresses are very high. Delamination can also develop from low-velocity impact [1]. Toughened resins can be used to reduce the initiation and growth of delaminations, but these are much more expensive than conventional resins. A second approach to improve delamination resistance is through mechanical means such as through-the-thickness stitching. Recent test results [1] show that stitched laminates, made using relatively inexpensive conventional fabric-resin combinations, offer damage tolerance equal to those using toughened matrix materials. To minimize damage from stitching, dry carbon fabrics were stitched with closely spaced heavy threads and then fabricated using resin transfer molding. However, stitching may cause fiber crimping or puncture damage that will degrade fatigue life.

This investigation evaluates the performance of stitched uniweave and toughened tape materials and addresses the effects of stitching on compressive fatigue life for open hole spec-

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TABLE 1—Composite properties for stitched AS4/3501-6 uniweave and unstitched IM7/8551-7 tape.

Composite Properties	AS4/3501-6 Stitched	IM7/8551-7 Unstitched
Carbon fiber areal weight, g m^{-2}	157	145
Total fiber areal weight, TFAW, g m^{-2}	171	145
Weight fractions from acid digestion	62.5	61.9
Carbon fiber density, ρ_f , g cm^{-3}	1.78	1.78
Resin density, ρ_r , g cm^{-3}	1.27	1.27
Composite density, ρ_c , g cm^{-3}	1.59	1.55

imens and postimpact compression specimens. To this end, a series of postimpact compressive fatigue and open hole compressive fatigue tests was conducted using specimens made from carbon-epoxy uniweave composite (AS4/3501-6) with stitching and a toughened carbon-epoxy tape (IM7/8551-7) without stitching. Results for the stitched material were compared with results for unstitched uniweave AS4/3501-6 [1].

Description of Materials

Toughened IM7/8551-7 Tape

The IM7/8551-7 tape material was laid up into a 48-ply quasi-isotropic laminate with a $[45^\circ/0^\circ/-45^\circ/90^\circ]_{6s}$ stacking sequence. The IM7 fibers are intermediate modulus carbon fibers and the 8551-7 is a toughened epoxy. The laminate was manufactured by Douglas Aircraft Company³ from tape prepreg using conventional layup techniques and cured in an autoclave according to manufacturer’s instructions. This material system was formulated to offer greater interlaminar fracture toughness, and hence more resistance to damage growth, than brittle carbon epoxies. Composite properties are given in Table 1. The unnotched static strengths and average specimen thickness are given in Table 2 for the IM7/8551-7 material.

Stitched AS4/3501-6 Uniweave

A brittle epoxy matrix composite system, AS4/3501-6, was selected for study because it is widely used in tape form. Also, the resin has a relatively low cost and accommodates the resin transfer molding process. The dry fabric was manufactured by Textile Technologies, Inc.³ The dry fabric consisted of 95% carbon fibers and 5% fiber glass fill yarn by weight. A 225-denier fiberglass yarn was woven in the fill direction to hold the fibers together while stitching. (Denier is a term used in the textile industry to designate the weight in grams of a 9000-m length of thread.) Figure 1 shows a specimen made of the uniweave material without stitching. The fill yarn is the light-colored tow, and the carbon is the dark-colored tow.

Individual layers of fibers were laid up to form a quasi-isotropic laminate with a $[45^\circ/0^\circ/-45^\circ/90^\circ]_{6s}$ stacking sequence. These layups were then lock-stitched (Fig. 2) with a heavy (3678 denier) fiberglass thread (S-2 Glass 449-1250 untwisted) at 3.2-mm ($\frac{1}{8}$ -in.) row spacing and 3.2 mm per stitch (eight stitches per inch). The stitching was 4% of the carbon weight. The dry carbon preforms were then impregnated with 3501-6 matrix resin by Douglas Aircraft Company using their patented resin transfer molding (RTM) process [1]. Composite properties are given in Table 1 for the stitched AS4/3501-6 uniweave material. The unnotched static

³ Use of manufacturers’ trade names does not constitute endorsement either expressed or implied by the National Aeronautics and Space Administration.
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TABLE 2—Average thickness and unnotched static compressive strengths for stitched AS4/3501-6 uniweave and unstitched IM7/8551-7 tape specimens.

Material System	Thickness, mm	Unnotched Strength, MPa
AS4/3501-6 stitched uniweave	9.22	472.5
IM7/8551-7 unstitched tape	7.45	620.0

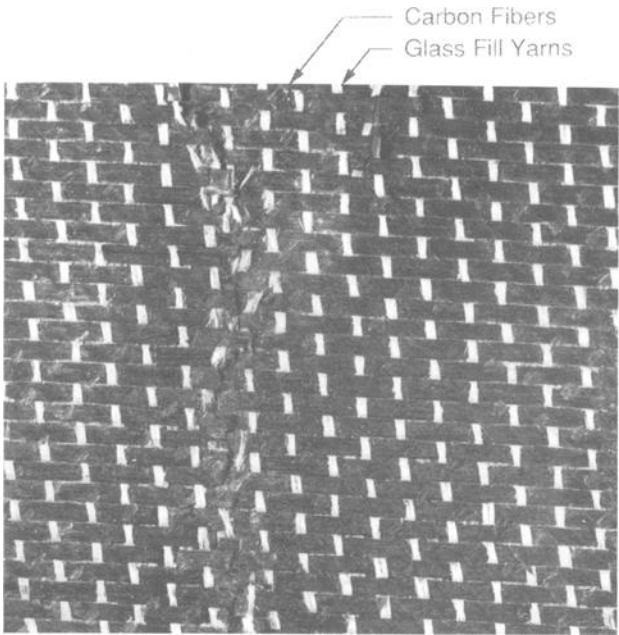


FIG. 1—Photograph of an unstitched uniweave specimen.

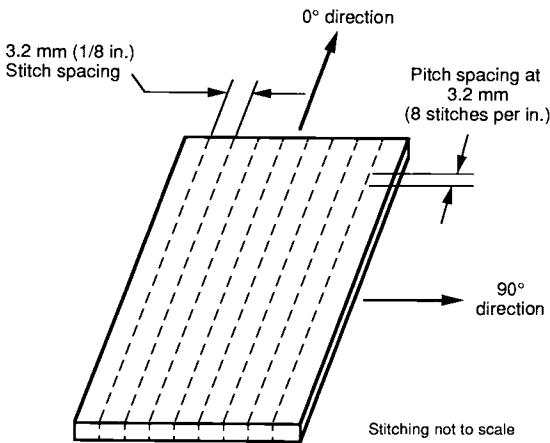


FIG. 2—AS4/3501-6 carbon-epoxy laminate with unidirectional stitching.

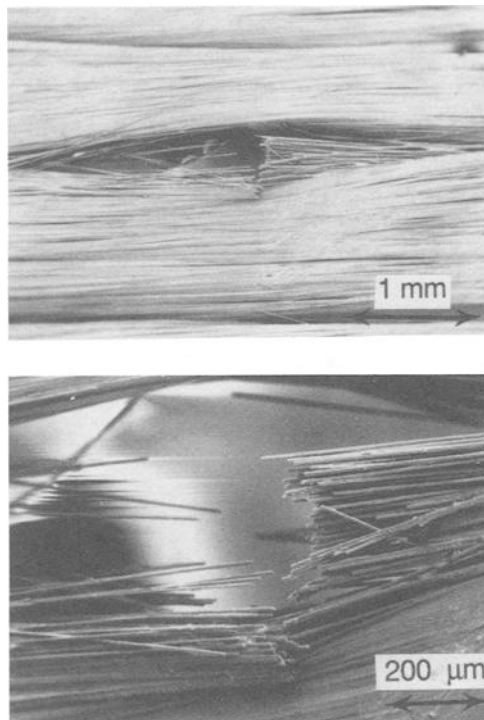


FIG. 3—Photomicrographs of a stitch site from AS4/3501-6 after pyrolysis: (top) magnification $\times 30$ and (bottom) magnification $\times 115$.

strength for the stitched uniweave composite is given in Table 2 along with the average specimen thickness. The stitched uniweave composite is about 20% thicker than the toughened tape composite because of the additional fill and stitch fibers. (An increase in fiber content causes an increase in matrix if the fiber weight fraction remains constant.)

Photomicrographs at two different magnifications ($\times 30$ and $\times 115$) are shown in Fig. 3 for a typical stitch site from an untested stitched composite specimen after the 3501-6 matrix material was removed by heating the specimen in air to 400°C (752°F) and holding the temperature for 3 h. Numerous fibers are broken by the stitching process. The crevasses at each stitch site were filled with resin during RTM. The damaged fibers are potential regions for damage initiation and growth, which may contribute to a reduction in static and fatigue strength.

Test Specimens

Open Hole Fatigue Specimens

Approximately 25-mm (1-in.) wide strips removed from the sides of the impact panels (Fig. 4) were used to produce the open hole fatigue specimens. The stitching in Fig. 4 is not to scale. The average specimen width was 22.1 mm (0.87 in.) for the AS4/3501-6 and 24.1 mm (0.95 in.) for the IM7/8551-7. The overall length of the specimen was 10 cm (4 in.), and the unsupported gage length (between grips) was 3.8 cm (1.5 in.). The load-bearing ends of each of the specimens were ground flat square and parallel. Each specimen had a 6.35-mm (0.25-in.) hole

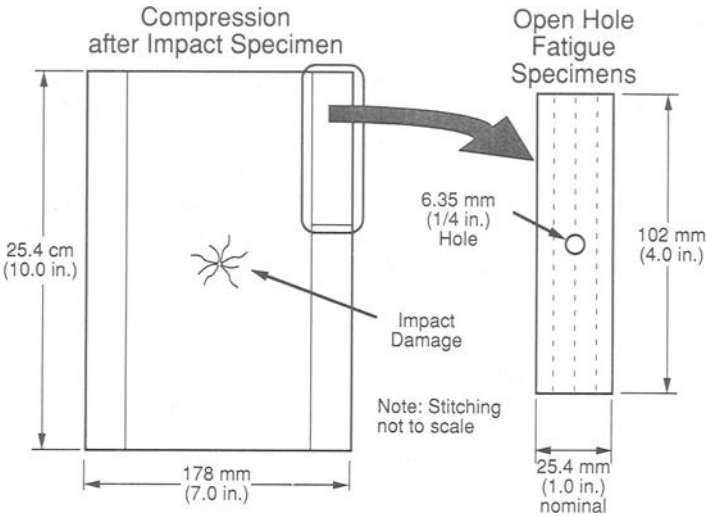


FIG. 4—Open hole specimen fabrication.

drilled through its center. These specimens were made narrower than the NASA open hole compression standard [2] because of a lack of material.

Postimpact Fatigue Specimens

A NASA standard 17.8- by 25.4-cm (7- by 10-in.) panel [2] was used for the impact study. After impact, the panels were machined to the NASA standard size 12.7 by 25.4 cm (5 by 10 in.) for compression testing. The load-bearing ends of each of the specimens were ground flat square and parallel. Figure 5 shows the size of the specimens.

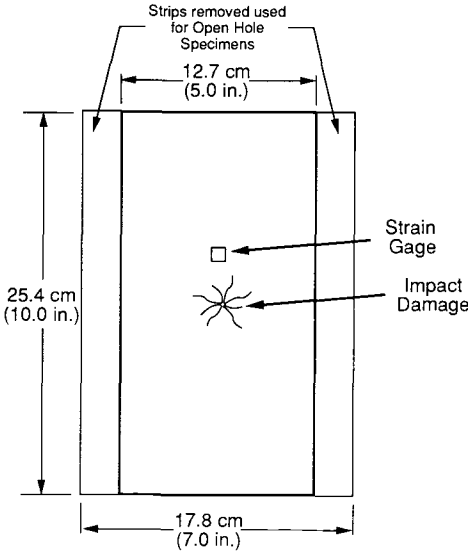


FIG. 5—Postimpact fatigue test specimen dimensions.

Test Equipment and Procedure

Open Hole Fatigue Tests

The test fixture used in the open hole fatigue study was originally developed for a compressive loading rate study [3]. This fixture was also used in a previous fatigue study on stitched materials by Lubowinski [4]. The fixture was designed to reduce misalignment, which can lead to premature failure. The fixture (Fig. 6) has alignment rods running on linear bearings at each of the four corners. The gripping surface has been serrated to transmit load, through shear, on the face of the specimen. The grips also contained end stops that were a few thousandths of an inch thinner than the specimens. These end stops eliminated any slipping of the specimen resulting from the high compressive loads.

Fatigue tests were run at room temperature in a closed-loop servohydraulic test machine at a frequency of 10 Hz. All specimens were tested in compressive-compressive fatigue with an *R* ratio (minimum/maximum) of 10. Typical failure sites were sectioned through, mounted in epoxy, and polished for examination under light microscopy.

Impact Test

The test fixture used for the impact portion of this study was originally developed by NASA Langley for compression after impact loading [2]. This fixture contained a 5- by 5-in. opening. A free-falling mass impacted the specimen. The impactor set-up and test fixture are illustrated in Fig. 7. The impactor consisted of three basic parts: a 51-mm (2-in.) diameter steel rod, an instrumented section of hollow rod, and a 12.7-mm (0.5-in.) diameter spherical tup. The instrumented section of hollow rod, or tup (Fig. 8), contained a piezoelectric accelerometer and four strain gages for measuring acceleration and impact force. The impact acceleration and impact force were recorded with a digital data acquisition system. The data were reduced on a desk-top PC using commercial software.

The 17.8- by 25.4-cm (7- by 10-in.) panels were mounted in the holder (see Fig. 7), and the bolts were torqued to 2.87 J (80 in.·lb). The instrumented impactor was centered above the panel at the required height to impart the desired impact energy per unit thickness. After the impactor struck the specimen, a piece of wood was quickly moved between the fixture and specimen to prevent the impactor from repeatedly hitting the panel.

An impact energy per unit thickness of 6672 J/m (1500 in.·lb/in.), which is an industry standard for evaluating thick, quasi-isotropic laminates, was used. The stitched AS4/3501-6

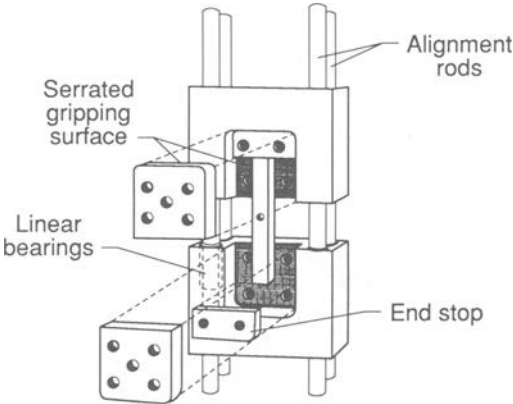


FIG. 6. Open-hole compressive fatigue test fixture.

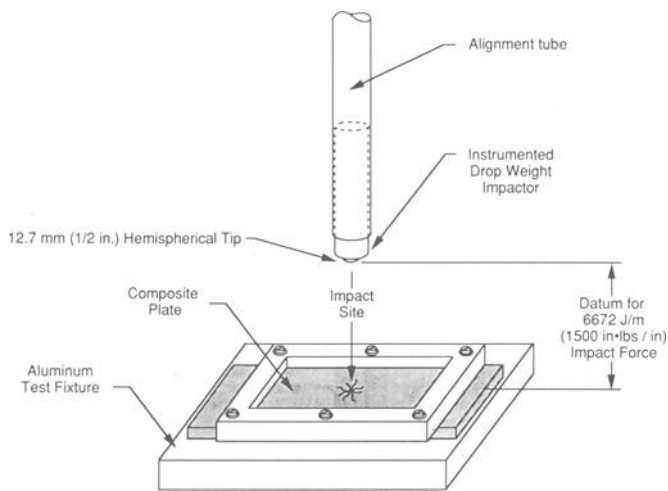


FIG. 7—Impacting of composite panels.

uniweave panels were impacted at 61.5 J (544 in.·lb), and the IM7/8551-7 tape panels were impacted at 50.2 J (444 in.·lb). These impact energies were chosen to yield the same energy per thickness, because of the difference in thickness between the IM7/8551-7 and the AS4/3501-6 materials.

Each panel was ultrasonically C-scanned before impact to ensure that the panels were of high quality and free from manufacturing defects that would lead to premature failure. C-scans after impact were used to determine the extent of the impact damage.

Postimpact Fatigue Test

Before each fatigue test was run, a single monotonic excursion was made to maximum load with load and strain recorded on an *x-y* chart recorder. The load was recorded directly from

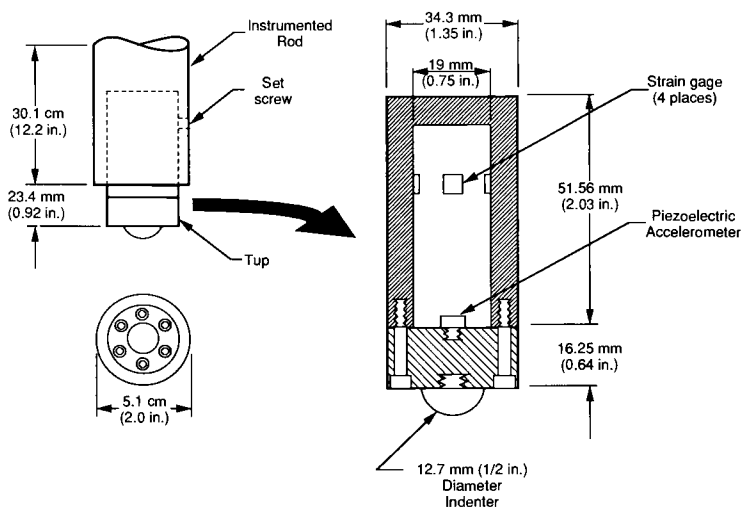


FIG. 8—Instrumented impactor tup.

TABLE 3—Damage area and elastic modulus for stitched AS4/3501-6 uniweave and unstitched IM7/8551-7 tape.

AS4/3501-6 Uniweave Stitched			IM7/8551-7 Tape Unstitched		
Panel ID	Damage Area, cm ²	Compressive Modulus, GPa	Panel ID	Damage Area, cm ²	Compressive Modulus, GPa
A01	18.85	49	B01	10.04	54
A02	20.14	43	B02	9.84	53
A03	18.56	42	B03	10.66	55
A04	22.12	46	B04	10.27	55
A05	18.71	51	B05	10.36	55
A06	22.32	50	B06	9.65	56
Standard deviation	1.72	3.7	Standard deviation	0.36	0.8
Mean	20.12	46.9	Mean	10.13	54.5

the test stand load cell, and strain was acquired from a surface-mounted strain gage, located immediately above the damage region, in the center of the test specimen (see Fig. 5). The slope of the load-versus-strain plot was used to obtain the elastic compressive modulus and is reported in Table 3. The fatigue tests were conducted at room temperature in a closed-loop servohydraulic test machine at a frequency of 5 Hz. All specimens were loaded in compressive-compressive fatigue with an *R* ratio (minimum/maximum) of 10. The fixture used for the fatigue test was originally developed for static compressive loading after impact [2]. Knife-edge supports prevent global buckling while the coupon is being loaded. The fixture is illustrated in Fig. 9.

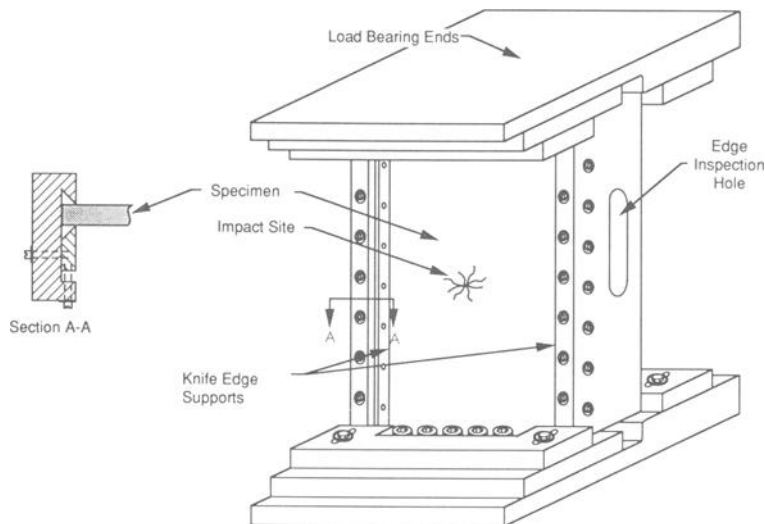


FIG. 9—Postimpact compression fatigue test fixture.

Discussion of Results

The carbon fiber areal weights were similar for each of the materials, but the proportion of carbon fibers was less for the stitched uniweave material than for the toughened tape material because of the presence of fill yarns. The stitched uniweave was 18% thicker on the average than the toughened tape, and, as a consequence, its weight was also greater. By comparing the materials on a stress basis alone, the excess thickness reduces the stress to the point that stitching appears to have a deleterious effect. To assess the effects of through-the-thickness stitching, it was necessary to compare the materials on an equal carbon content basis (equal load per unit width) as well as on an equal stress basis.

Open Hole Fatigue

Maximum compressive load per unit width, N_x , is plotted against cycles to failure (log scale) in Fig. 10 for the AS4/3501-6 uniweave composite in stitched and unstitched configurations and for the IM7/8551-7 tape composite. Linear least squares regression fits to the data are also plotted. The unstitched AS4/3501-6 uniweave specimens are made from the same woven preform as the stitched specimens and fabricated using the same RTM process [1]. The thickness of the unstitched uniweave specimens was 7.6 mm as compared to 9.22 mm for the stitched uniweave and 7.45 mm for the toughened tape (Table 2). Figure 10 shows that all three of the material configurations experience a reduction in compressive strength with constant amplitude compressive fatigue cycles of between 35 and 36% at 10^6 cycles. The fatigue lives are about the same for all of the materials tested, indicating that stitching the material has no deleterious effect, even though stitching may cause additional crimping and results in broken figures (see Fig. 3).

Test data for the open hole fatigue specimens of the stitched and unstitched AS4/3501-6 uniweave and the IM7/8551-7 tape composites are given in Table 4 as gross compressive stress, $\sigma_{\max}$, versus cycles to failure, where the gross compressive stress is

$$|\sigma_{\max}| = \frac{P}{W \cdot t} \tag{1}$$

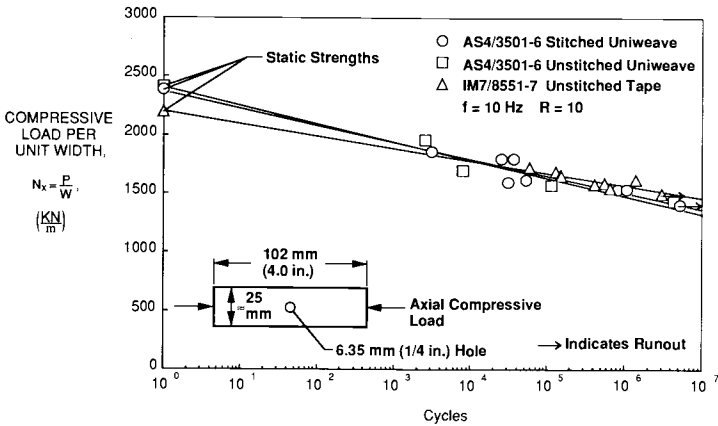


FIG. 10—Load per unit width versus cycles to failure for open hole compression specimens.

TABLE 4—Open hole static strengths and compressive fatigue data for stitched and unstitched AS4/3501-6 uniweave and unstitched IM7/8551-7 tape.

AS4/3501-6 Stitched		AS4/3501-6 Unstitched [1]		IM7/8551-7 Unstitched	
Compressive Stress, MPa	Cycles	Compressive Stress, MPa	Cycles	Compressive Stress, MPa	Cycles
258.96	1	317.17	1	298.00	1
202.07	3160	258.62	2506	233.52	5.77E4
195.52	2.51E4	224.14	7860	228.21	1.27E5
195.52	3.63E4	206.90	1.12E5	223.79	1.51E5
175.79	5.22E4	189.66	4.37E6	223.52	1.45E5
173.10	3.08E4	...	...	214.76	5.46E5
166.97	1.08E6	...	...	218.97	1.39E6
166.83	8.22E5	...	...	213.66	4.16E5
159.03	5.00E6 ^a	...	...	209.66	6.56E5
...	...	...	...	203.52	3.00E6 ^a

^aSignifies runout.

where

- P = maximum compressive load,
- W = specimen width, and
- t = specimen thickness.

The data from Ref 1 for the unstitched AS4/3501-6 uniweave are included for convenience of the reader.

Gross compressive stress versus the number of cycles to failure (log scale) is plotted in Fig. 11 for the AS4/3501-6 uniweave in stitched and unstitched configurations, and the IM7/8551-7 tape material. The individual composite densities varied less than 3% so this figure is equivalent to a plot of the specific stress. Therefore, the higher the curve position on this figure, the lighter and stronger the composite. The IM7/8551-7 has the greatest fatigue life per unit weight but only slightly better than the unstitched AS4/3501-6. The greater thickness of the stitched AS4/3501-6 uniweave caused its fatigue life to fall well below the other materials. Thus, it appears that the increased thickness and weight from stitching are much more of a penalty in these tests than crimped fibers or puncture damage from stitching. The 76.2-mm-wide specimens in Ref 1 give somewhat larger static strength for the stitched fabric and the toughened tape composite than the 25.4-mm-wide open hole specimens tested in this study. The fatigue strengths may also be lower for the 25.4-mm-wide specimens.

Fatigue Damage Progression—Open Hole

Edge-replicating techniques using a substrate of cellulose acetate and a reagent-grade acetone were used to produce replicas. The replicas were taken on the outside edge of the test specimen, adjacent to the hole, at various cycle counts during fatigue loading. The holes were too small to make replicates in them. These edge replicates were then examined under light microscopy for evidence of damage. Visible observations on the surface indicated that damage tended to initiate slightly earlier in the hole than at the edge of the specimen but tended to

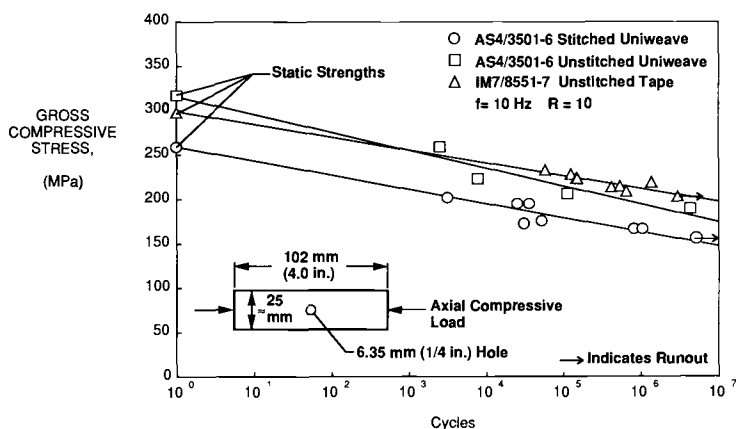


FIG. 11—Gross compressive stress versus cycles to failure for open hole compression specimens.

progress across the width of the specimen at a similar rate. In fact, the damage tended to join midway between the edge and the hole. The damage remained in a narrow region across the net section of the specimen and did not grow along the length.

The photomicrographs in Figs. 12 through 15 show the progression of damage through the thickness, at the portion of the straight edge adjacent to the hole, during fatigue loading of the stitched AS4/3501-6 uniweave. Figure 12 is a photomicrograph of the edge at 210K cycles. There is no evidence of damage at this point in the load history. Figure 13 shows that damage to a depth of 2 mm has initiated at the surface ply free edge within the next 90K cycles. The damage consisted of matrix cracking and delamination. Figure 14 shows that this damage remained mostly unchanged from 300 to 650K cycles, tending to grow only in the vicinity of the surface plies. Although the damage did spread across the width, it did not spread through

Surface Ply

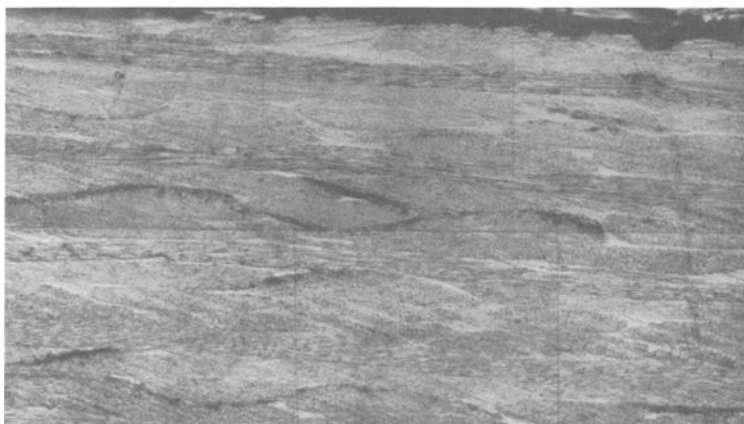


FIG. 12—Photomicrograph of stitched AS4/3501-6 after 210 000 cycles. $R = 10$, 167-MPa maximum compressive stress, 822 000 total cycles to failure.

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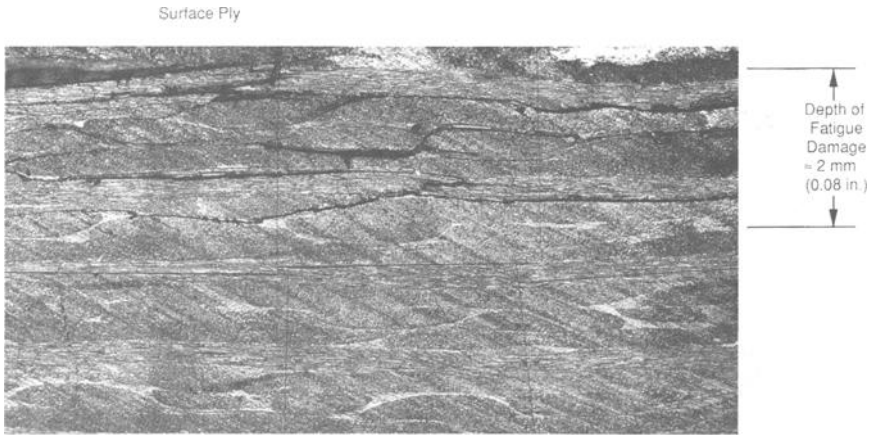


FIG. 13—Photomicrograph of stitched AS4/3501-6 after 300 000 cycles. $R = 10$, 167-MPa maximum compressive stress, 822 000 total cycles to failure.

the thickness until near final rupture. Carbon-epoxy materials typically fail under monotonic compressive loading as a result of matrix cracking and delaminations as well as some fiber breakage [5]. In this material, the delaminations were localized at the hole region, and final failure was by shear buckling of the fibers. Others have shown that, although the delaminations start earlier in the fatigue life, they progress at a slow rate in stitched materials [4]. The stitching apparently retards delamination growth by suppressing the opening mode. At some point near failure, matrix cracking and delaminations appear at ply levels well below the surface. This phenomenon can be seen in Fig. 15, a photomicrograph taken at 800K cycles, where failure is only 22K cycles away. The progression of failure in the stitched material is typical of that



FIG. 14—Photomicrograph of stitched AS4/3501-6 after 650 000 cycles. $R = 10$, 167-MPa maximum compressive stress, 822 000 total cycles to failure.

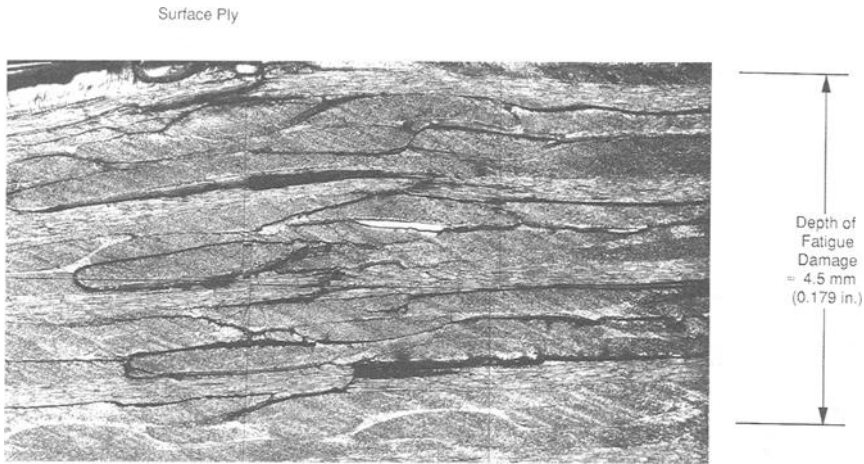


FIG. 15—Photomicrograph of stitched AS4/3501-6 after 800 000 cycles. $R = 10$, 167-MPa maximum compressive stress, 822 000 total cycles to failure.

obtained in tape materials of similar constituents. Thus, stitching does not alter the failure mode of these materials, it only slows its progression.

Impact Damage

Plots of impact force versus time for both the stitched AS4/3501-6 uniweave and the IM7/8551-7 tape are shown in Figs. 16 and 17. The peak impact force was approximately 2041 kg (4500 lbs) for both materials. The force-time signals for all the impact tests were consistent. The impactor rebounded on the first collision and was out of the way before the plate rebounded. The high-frequency oscillations throughout the load-time histories are mostly ringing of the impactor and vibrations of the specimens.

After impact, each panel was ultrasonically C-scanned to examine the extent of damage. Typical C-scan results for both materials are shown in Figs. 18 and 19. The damaged area, seen as the white region in the center of these photographs, was measured with the use of a planim-

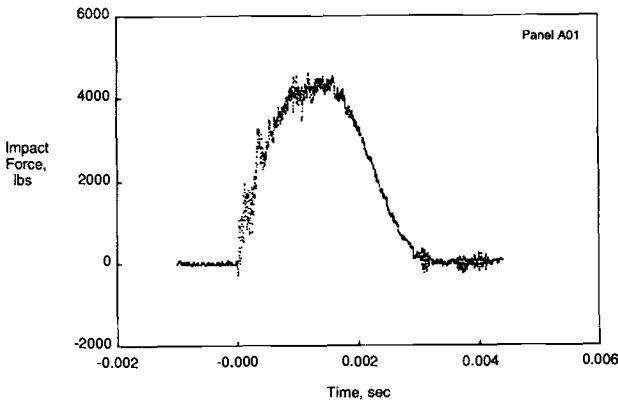


FIG. 16—Impact force versus time for stitched AS4/3501-6 (544 in. · lb).

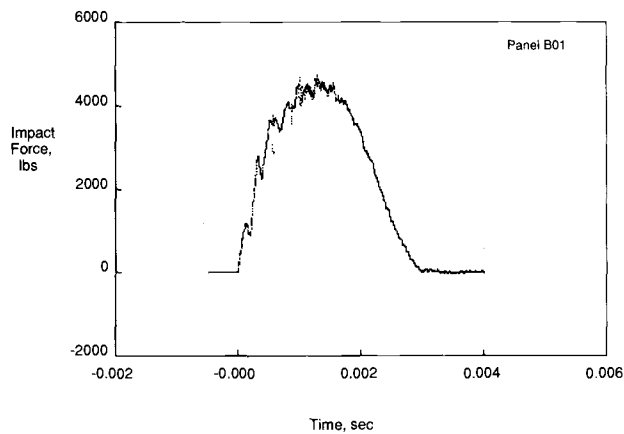


FIG. 17—Impact force versus time for stitched IM7/8551-7 (444 in. · lb).

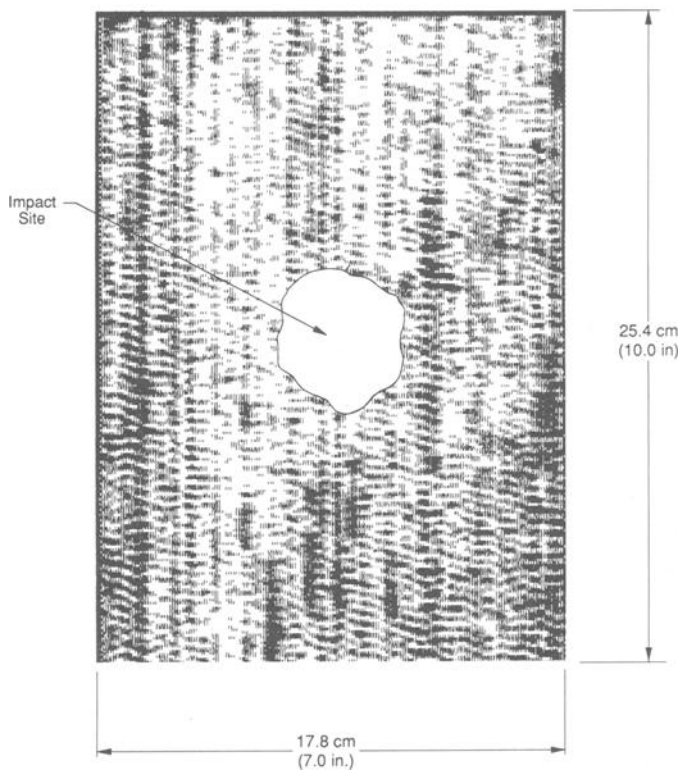


FIG. 18—C-scan of stitched AS4/3501-6 after impact.

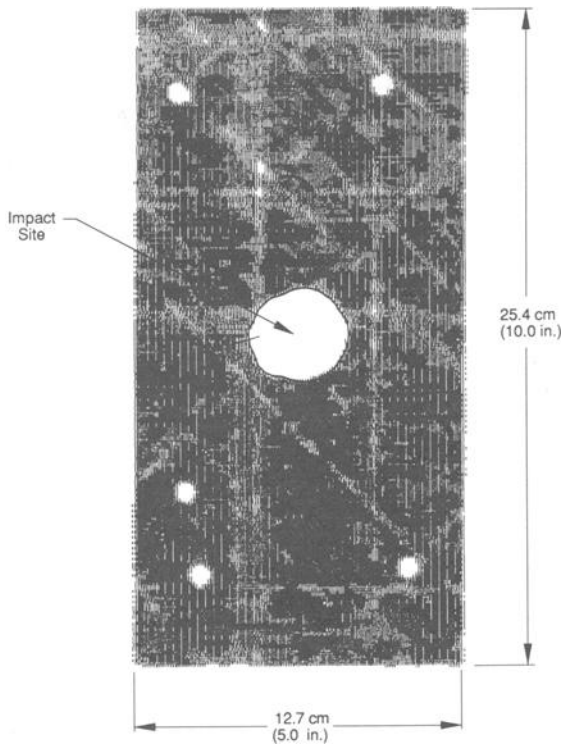


FIG. 19—C-scan of IM7/8551-7 after impact.

eter (see Table 3). The damaged region in the AS4/3501-6 was about twice as large as in the IM7/8551-7. Stitching limits the damage area in AS4/3501-6 uniweave composite [6], but the toughened resin tape composite had an even smaller damaged area. The standard deviation was calculated for each of the impact area data sets. It showed a variation of less than 9% for the AS4/3501-6 and less than 4% for the IM7/8551-7. This low standard deviation indicates that the impacts were consistent and repeatable.

Postimpact Fatigue

Test data for individual specimens of the stitched AS4/3501-6 uniweave and the toughened IM7/8551-7 tape laminates are given in Table 5. To assess the effects of damage due to stitching, the materials were compared on an equal carbon fiber areal weight basis. Maximum compressive load per unit width is plotted against cycles to failure (log scale) in Fig. 20 for the stitched AS4/3501-6 uniweave and the IM7/8551-7 tape composite. Linear least squares regression fits to the data are also plotted. No data were available for the unstitched AS4/3501-6 uniweave composite. Figure 20 shows that both the stitched material and the toughened system experience a reduction in compression strength with constant amplitude compressive fatigue cycles of between 33 and 34% at 10^6 cycles. The compressive fatigue lives of the stitched AS4/3501-6 are significantly longer than the IM7/8551-7. The static strengths are also larger.

Gross compressive stress is plotted against cycles to failure (log scale) in Fig. 21 for the

TABLE 5—Post-impact static strengths and compressive fatigue data for stitched AS4/3501-6 uniweave and toughened IM7/8551-7 tape.

AS4/3501-6 Stitched		IM7/8551-7 Unstitched	
Stress, MPa	Cycles	Stress, MPa	Cycles
343.45	1	334.41	1
250.34	2.92E4	250.34	7050
250.34	1.26E5	238.97	4.81E4
238.97	1.29E4	227.59	3.57E4
238.97	1.09E5	227.59	7.79E4
227.59	1.21E6	227.59	3.01E5
227.59	1.37E6	...	...

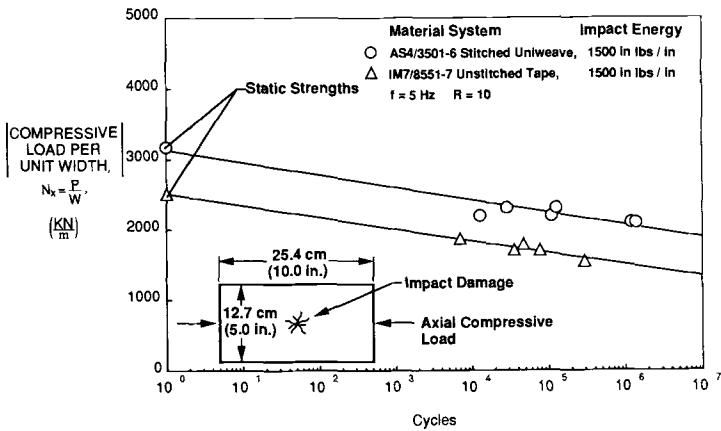


FIG. 20—Load per unit width versus cycles to failure for specimens after impact.

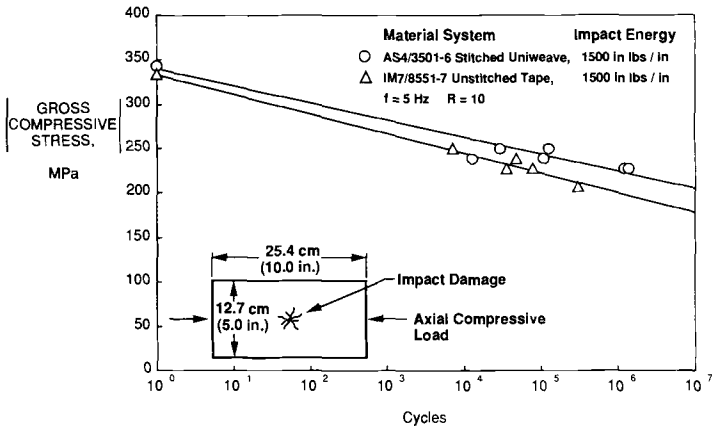


FIG. 21—Gross compressive stress versus cycles to failure for specimens after impact.

stitched AS4/3501-6 uniweave and the IM7/8551-7 tape material. Linear least squares regression fits to the data are also plotted. Here, as in Fig. 11, the higher the curve position on this figure, the lighter the material. The static strengths of the IM7/8551-7 and the stitched AS4/3501-6 are similar. However, the fatigue life of the stitched material is slightly better, even though it is thicker. Thus, the stitched material does not suffer a weight penalty. Recalling that the additional thickness caused the fatigue life of the open hole specimens to fall well below the toughened tape composite (Fig. 11), Fig. 21 indicates that the effect of additional thickness in the stitched AS4/3501-6 material was not as detrimental for postimpact damage life as it is for open hole fatigue life.

Conclusions

An investigation was conducted to evaluate the fatigue performance of a stitched uniweave fabric composite and a more expensive toughened resin tape composite. The effects of stitching on the compressive fatigue life of specimens with open holes and impact damage were also examined for the uniweave fabric.

The stitched uniweave composites were thicker than the tape composite because of the additional fill and stitch yarns. For open hole specimens, compared on a stress basis, the fatigue lives of the stitched uniweave composite were reduced in proportion to the increase in thickness. Hence, there was no apparent increase in life from stitching. For open hole specimens, compared on equal carbon fiber areal weight basis, the compressive static and fatigue strengths of the stitched and unstitched fabric composite and the toughened tape composite were about equal. Thus, comparisons based on an equal carbon content indicate that the crimp and puncture damage from the stitching had no net effect on the compression fatigue strength.

For impact damage, compared on equal carbon fiber areal weight and impact energy level, the compressive static and fatigue strengths of the stitched uniweave composite were greater than those of the toughened tape composite. The toughened tape composite was more resistant to initial impact, but the stitched uniweave composite was considerably more tolerant to damage. Thus, the impacted uniweave composite did not suffer a weight penalty, despite the increase in thickness. Therefore, it appears that the increase in thickness from stitching is more of a penalty than crimped fibers or puncture damage from stitching.

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A Method for Evaluating the High Strain Rate Compressive Properties of Composite Materials

REFERENCE: Montiel, D. M. and Williams, C. J., "A Method for Evaluating the High Strain Rate Compressive Properties of Composite Materials," *Composite Materials: Testing and Design (Tenth Volume)*, ASTM STP 1120, Glenn C. Grimes, Ed., American Society for Testing and Materials, Philadelphia, 1992, pp. 54–65.

ABSTRACT: A test apparatus and method developed to obtain high strain rate compressive mechanical properties of various fiber-reinforced composite materials is described. This direct compressive method uses a drop tower to impart a load at dynamic rates to the test fixture. Uniform specimen loading is accomplished through the use of aligned guiderails to constrain all but the vertical motion of a free-sliding impactor. Specimen endcaps, specific to thick-section testing, are implemented to prevent premature brooming failure. A piezoelectric transducer and aluminum absorbers are some of the refinements introduced to allow acquisition of stress and strain data free of distortion. Strain rates on the order of 8 s^{-1} have been achieved, and higher rates appear possible. Material property data were obtained on a series of AS4 graphite/PEEK thermoplastic $[\text{O}_2/\text{90}]_s$ composite laminates. The results of this first study indicate that at high strain rate loading, the strength of this material increases 42% over static values, while the strain to failure increases by 25%. There also appears to be a small effect of increased strain rate on the initial and secant moduli for this material.

KEYWORDS: composites, high strain rate, compressive properties, fiber-reinforced composites, loading

Composites have shown great promise as high-strength structural materials for a variety of purposes because of the potential benefits of decreased weight and increased stiffness. However, only a limited body of knowledge exists of the high strain rate properties and failure behavior of these materials to support dependable and efficient design for many applications. One such application is the response of composites to shock loading. Analytical methods have been used to simulate the structural response of a composite cylinder during a dynamic shock event. However, for these models to be accurate, realistic high strain rate material property input is needed in the analysis. Static data are not adequate for this purpose, because modulus, strength, strain-to-failure, and failure modes of composites can vary with strain rate [1,2].

Unfortunately, the high strain rate behavior of the thick-section composites used in this application is a relatively unexplored area. Dynamic testing of these materials in general has

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been very limited, and the results found in the available literature are inconsistent because of limited test techniques and specialized specimen geometries. The lack of dependable high strain rate material property data has been a major limitation in the analysis of composite structures subjected to shock loading. This work was initiated to develop the experimental techniques required to assess the high strain rate compressive properties of proposed thick-section composite material systems. Specifically, a test methodology for high strain rate compression testing of end-loaded thick-section composite laminate coupons has been developed.

The definition of strain rate ranges is generally arbitrary because the range is usually defined by the material response. In using the term *high strain rate*, we have followed the guidance of Rotem and Lifshitz [3], who categorized the high strain rate regime for composites as being in the range of 5 to 30 s⁻¹.

Experimental Procedure

Test Method and Supporting Apparatus

Earlier in-house investigations indicated that static thick-section laminate compression test fixtures could not be readily adapted for service in a dynamic loading environment. The first phase of this work was to design a new compressive apparatus, the purpose being to impact dynamically an end-loaded coupon, where load is introduced by direct bearing on the end of a short, rectangular specimen of appreciable thickness (greater than 2.54 mm thick) and unsupported gage length. The new apparatus is pictured in Fig. 1. The fixture consists of a T-

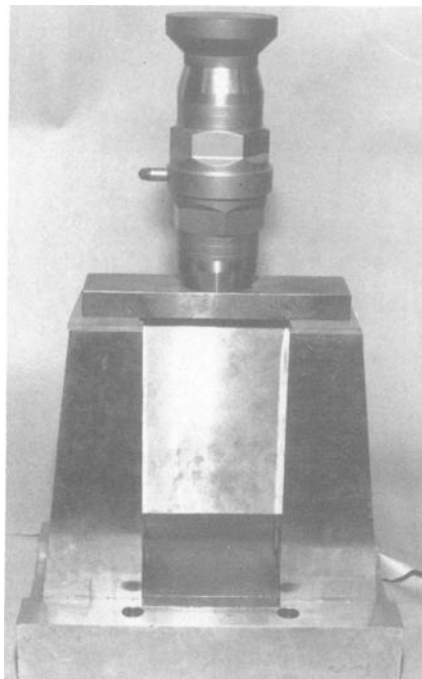


FIG. 1 High-strain rate compression test fixture.

shaped impactor, aligned and directed by two vertical guiderails, one on each side. The guiderails are bolted into a channel contained in a steel base perpendicular to the guiderails; this serves to align the guiderails both perpendicularly to the base and laterally with one another. The lateral distance of separation between the guiderails can be adjusted to accommodate free sliding of the impactor. The impactor was machined from 4340 steel, then hardened to 44 Rockwell Hardness (C Scale) (HRC). A hardened 4340 steel bottom plate is seated between the guiderails to provide a flat surface, parallel to the impactor bottom face, on which to place the specimen. Close tolerances were specified for flatness and perpendicularity on the contact faces to minimize specimen out-of-plane loading.

A drop tower assembly was used to achieve a loading duration within an order of magnitude of that achieved during the explosion tear test, which was designed to be characteristic of the loading response of a metallic structure to dynamic loading [4]. The tower uses a free-falling crosshead, guided by a large load frame, to impart a load to the test fixture. The test fixture is shown installed in the drop tower in Fig. 2. The tower tup directly contacts the top of the impactor; a vertical stud on top of the impactor is threaded to accommodate a piezoelectric load cell, which records the load feedback. A small, pyramidal absorber made of T6061 aluminum in the fully annealed condition was positioned between the load cell and the drop tower tup. This absorber dampens the spurious noise in the load and strain histories caused by apparatus vibrations and steel-on-steel contact; the pyramid shape serves to distribute smoothly and evenly the load to the T-shaped impactor. This technique has been employed

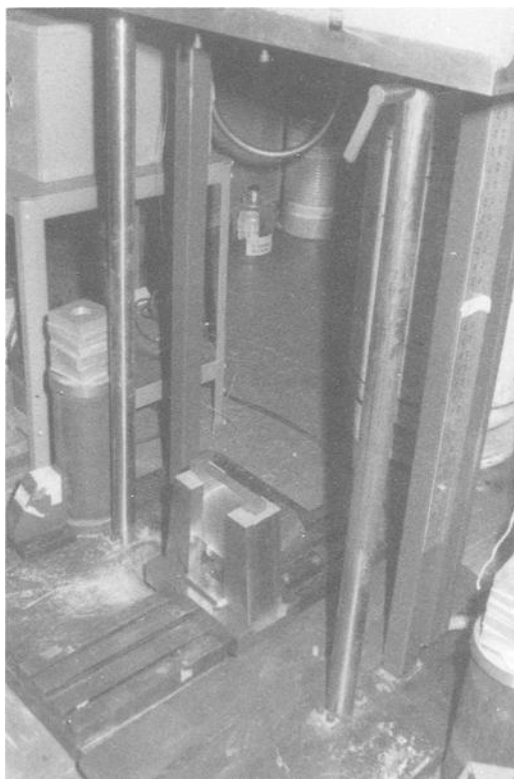


FIG. 2. Test fixture as installed in the drop tower assembly.

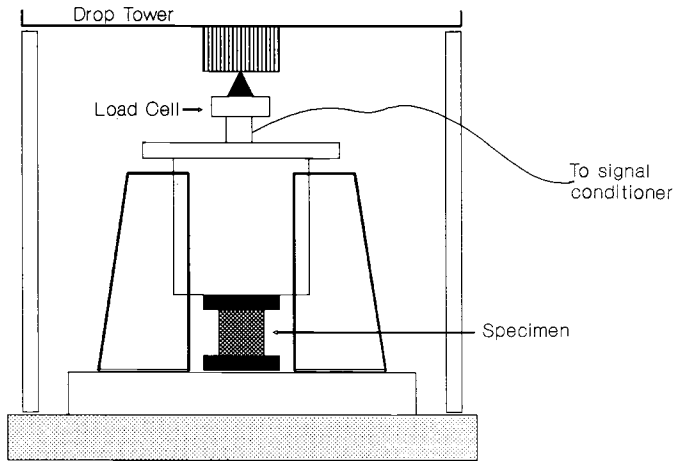


FIG. 3—Schematic of dynamic compressive fixture.

by Joyce and Hackett [5] using different absorber configurations. A schematic of the test fixture showing the aluminum absorber in place is given as Fig. 3. The crosshead was dropped from 0.305 m (2 ft) and was arrested by rigid stops after the specimen was impacted and displaced to failure. The crosshead mass used in these experiments was 1760 kg. Strain data were obtained from strain gages mounted directly on the specimen. The data from the load cell and the specimen strain data were recorded on a Nicolet 4094 series digital oscilloscope, then transferred to an IBM PC for data reduction and analysis. The strain rate was attained by differentiation of the strain-time curve. The tup velocity at impact was recorded using an optical flag system.

Specimen Design

The specimen used in this study was modeled after the thick-section test coupon used at NASA Langley Research Center for static compression testing [6]. The specimen size is nominally 6.4 mm (0.25 in.) thick by 31.8 mm (1.25 in.) wide. The coupon length is 44.4 mm (1.75 in.), which includes a 25.4-mm (1.0-in.) unsupported gage section and 9.5-mm (0.375-in.) grip lengths on either end. These dimensions were considered appropriate for the initial material study; however, the fixture can accommodate specimens through a wide range of sizes. Specimen ends are ground flat and parallel to within 0.0254 mm (0.001 in.) to ensure uniform load introduction to the specimen.

Specimen endcaps are used to suppress end-initiated brooming failure, which often occurs in end-loaded specimens with large percentages of fibers oriented in the load (0°) direction. Prevention of brooming in coupon specimens allows more direct comparison between coupon data and cylindrical specimen data, because the modeled "zero direction" of a cylinder is circumferential and, therefore, is not susceptible to end-initiated failure. The specimen is gripped using the reusable, rectangular endcaps, as illustrated in Fig. 4. The specimen ends are recessed into the hardened steel caps to a depth of 9.5 mm (0.375 in.), then bonded using a room temperature cure epoxy adhesive to provide constraint. A similar gripping concept has been used successfully with cylindrical metal matrix composite specimens by Lamothe and Nunes [7] and with glass/polyester specimens by Han [8]. The specimen grips are separable from the fixture itself and can be used in static testing as well. Each endcap consists of two pieces, a

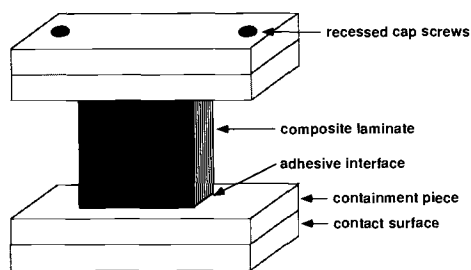


FIG. 4—Schematic of specimen geometry and endcap design.

containment piece and a contact surface (see Fig. 3). The first purpose of this design is accurate specimen alignment. The specimen is first bonded into the containment piece with the epoxy adhesive. During the epoxy cure cycle, the specimen end is held against the flat contact surface. After curing, the containment piece and contact surface can be unbolted, allowing visual examination of the specimen end to make sure it is flush and parallel to the endcap. The other advantage of this design is ease of specimen removal. After testing, the contact surface can be removed from the endcap, facilitating removal of the specimen for postfailure examination without incurring further damage.

The specimens were instrumented with longitudinal strain gages to measure longitudinal strain and to detect any out-of-plane deflection that would occur from global specimen instabilities. A 6.35-mm (0.25-in.) longitudinal gage was applied in the center of the gage length on each side of the specimen. The strain gages were applied with M-bond AE-10 adhesive to prevent premature debonding.

Material

The material investigated in this initial study was an AS4 graphite/APC-2 poly(etheretherketone; PEEK) thermoplastic composite. Carbon-reinforced PEEK is a high-performance composite that is potentially advantageous for many applications compared to carbon/epoxy systems. The toughness of the material gives good impact resistance, and its flammability, toxicity, and smoke density characteristics are better than those of epoxy systems.

The material was supplied as a hot press-molded plate by Fiberite Corporation. The plate was fabricated from unidirectional prepreg tape and supplied as a 48-ply (6.4-mm thickness) continuous fiber-reinforced laminate of $[0_2/90]_{8s}$ orientation. Fiber volume fraction was calculated to be 61% from a resin content of 32% by weight. Void content was specified by the manufacturer to be less than 1%. An ultrasonic C-scan inspection was performed, which indicated no defects.

Results and Discussion

Mechanical Property Data

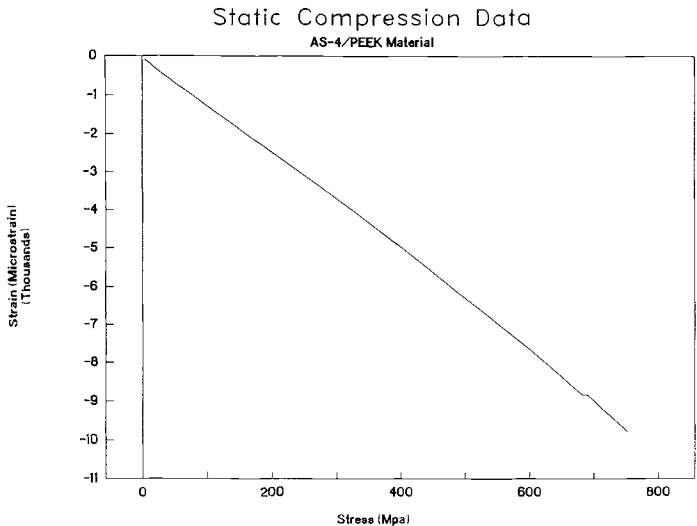
After completing initial proof of concept tests both statically and dynamically, a series of five specimen tests were conducted on the AS4/PEEK specimens for both the static and high strain rate conditions. The original data were then smoothed using a three-point averaging technique before data analysis. The static test results obtained using the endcap grips were comparable to static results obtained for the same material using another thick-section, end-loaded grip method [9]. Data for the static tests are given in Table 1. The average static com-

TABLE 1—Static compression test data for AS4/PEEK Material $[0_2/90]_{8s}$ orientation.

Specimen ID	Area, mm ²	Ultimate Compression Strength, Mpa	Maximum Strain (microstrain)		Initial Modulus, Gpa	Secant Modulus, Gpa
			[V1]	[V2]		
AC-1	201.6	841	11 490	10 900	83.7	78.0
AC-2	199.1	883	11 800	11 000	85.0	79.9
AC-3	198.7	717	9 500	9 400	81.7	76.0
AC-4	196.6	758	9 794	10 070	82.8	77.8
AC-6	200.5	793	9 200	10 100	86.5	82.2
Average:	199.3	798.4	10 325		83.8	78.8

pressive strength obtained for this material was 800 MPa (116 ksi). The average initial modulus, as calculated in accordance with the Suppliers of Advanced Composite Materials Association (SACMA) recommended test method SRM 1-88 [10], was 83.8 GPa (12.7 million psi). The average secant modulus, calculated between strain values of 0 and 9000 microstrain, was 78.8 GPa (11.42 million psi). These values show a 6% change between the initial and secant moduli, which represents numerically a small nonlinearity in the stress-strain response for this material and orientation. A typical stress-strain curve for these tests is shown in Fig. 5. The average strain-to-failure for these tests was 1.03%.

High strain rate test results obtained with the proposed fixture are shown in Table 2. Figure 6 shows typical load cell and strain gage responses for these specimens. It should be noted that the strains track very well and that peak strains and peak load occur almost simultaneously. Figure 7 illustrates the degree of alignment achieved during testing; the two signals plotted are from back-to-back vertical strain gages mounted oppositely on a representative specimen. The lack of bending indicates a good degree of alignment between the drop tower tup, fixture, and specimen. No significant bending was observed in any of the tests run. The stress-versus-strain



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FIG. 5—Static stress-strain curve.

TABLE 2—High strain rate test data for AS4/PEEK material ([0₂/90]_{8s} orientation).

Specimen ID	Area, mm ²	Ultimate Compression Strength, Mpa	Maximum Strain (microstrain)		Initial Modulus, Gpa	Secant Modulus, Gpa	Strain Rate, sec ⁻¹
			[V1]	[V2]			
AC-7	198.7	1186	13 044	12 150	94.6	87.0	7.4
AC-8	195.6	1145	12 817	12 938	92.2	90.8	7.8
AC-9	196.5	1103	13 968	...	...	...	8.1
AC-10	194.2	1099	12 999	10 878	91.9	89.8	7.8
AC-11	195.9	1138	14 120	14 029	83.7	81.8	8.2
Average:	196.2	1134	12 994		90.6	87.4	7.9

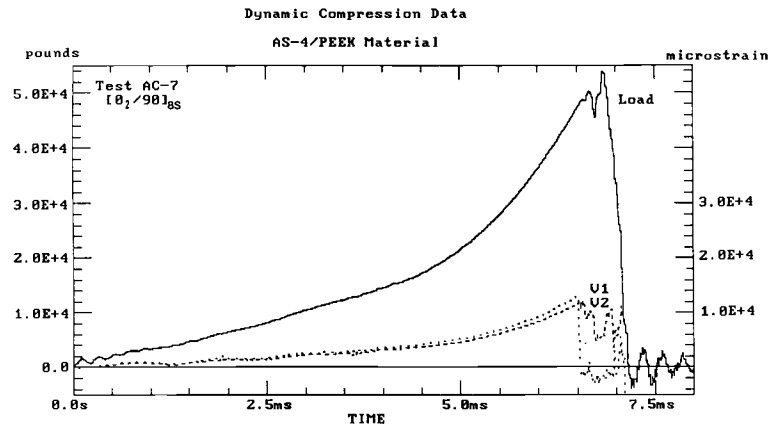


FIG. 6—Load and strain histories.

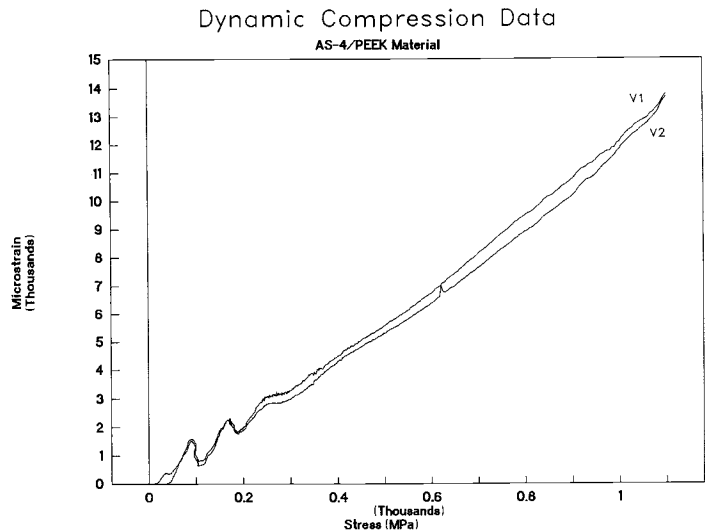


FIG. 7—Back-to-back longitudinal gage signals.

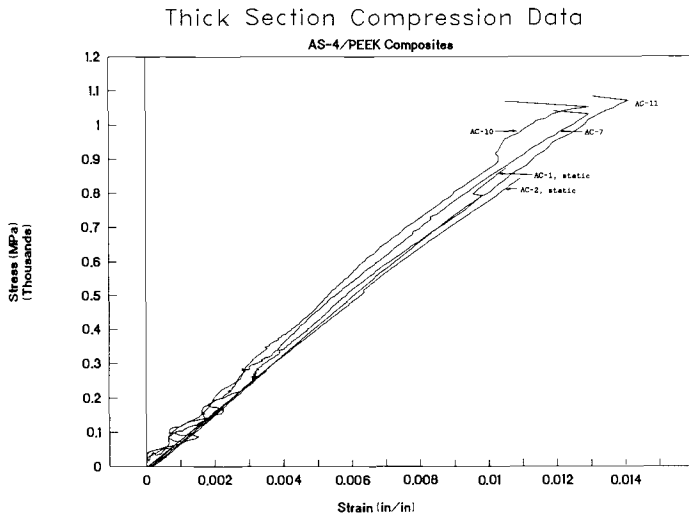


FIG. 8—Static and dynamic compression stress-strain plots.

traces for three tests are shown in Fig. 8, with two traces from static tests also plotted for comparison. Because absorbers were used during testing, the initial part of the strain-time curve was not truly indicative of the effective strain rate seen by the specimen. The strain rate appeared, however, to reach a constant rate at a load value of approximately 50% of ultimate load to failure for all tests. The strain rates for all tests, as reported herein, were obtained by differentiating the strain-time curve between the strain values corresponding to 50% of ultimate load, and just before failure. The mean strain rate attained by this method for these tests was 7.9 s^{-1} . The average dynamic compressive strength for the AS4/PEEK material at this rate was measured to be 1134 MPa (164 ksi). This represents a 42% increase over its measured static strength. The average measured dynamic ultimate strain of 1.30% for this material was also higher than the statically measured value and represents a 25% increase. The high strain rate initial modulus, also calculated according to SACMA method SRM 1-88, averaged 90.6 GPa (13.0 million psi) and was fairly reproducible for these specimens. This average initial modulus measured dynamically was approximately 7% higher than the average static value, showing a small effect of increased strain rates of this magnitude on the initial modulus of this material.

The average high strain rate secant modulus was 87.4 Gpa, which is 11% higher than the average static secant modulus. These comparisons indicate that the high strain rate modulus curves of this laminate exhibit less nonlinearity at higher strain values than the static modulus curves. The change in secant modulus is also reflected in the larger proportional increase shown in laminate dynamic strength (42%) than in laminate dynamic strain (25%), when comparing these values to those measured statically.

Failure Mode

Different mechanisms often govern the deformation behavior of materials within different strain rate regimes [11]. For this reason, the static and dynamic specimens were examined and compared for differences in failure behavior. The failure modes of specimens tested at static

and high strain rates turn out to be similar for this study, although greater damage was incurred in the high strain rate specimens. This is due, in part, to the continued travel of the impactor after specimen failure. The drop tower head was stopped as soon as possible after coupon failure to avoid crushing the specimen; however, the allowed travel distance was, by necessity, larger than the specimen strain-to-failure distance. A photograph of a typical failed specimen is shown in Fig. 9. Both static and dynamic specimens exhibited a regular delamination pattern as well as a crush zone. The delaminations occurred every three plies at one side of the 0 to 90° interface (Fig. 10). Crush zones were present close to the endcaps at one or both ends; kink band/shear failures can be observed in these regions for both static and dynamic specimens (Fig. 11).

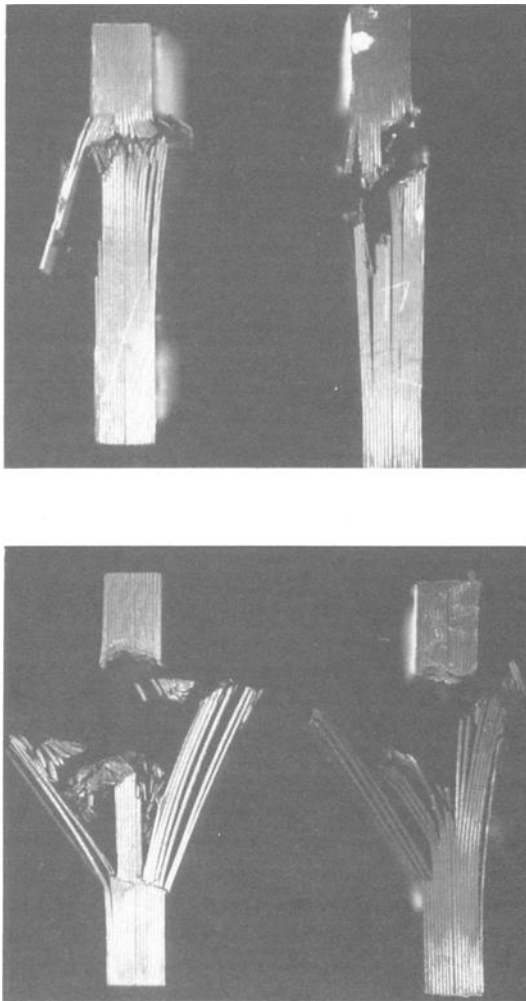


FIG. 9—Failed static (top) and dynamic (bottom) specimens.

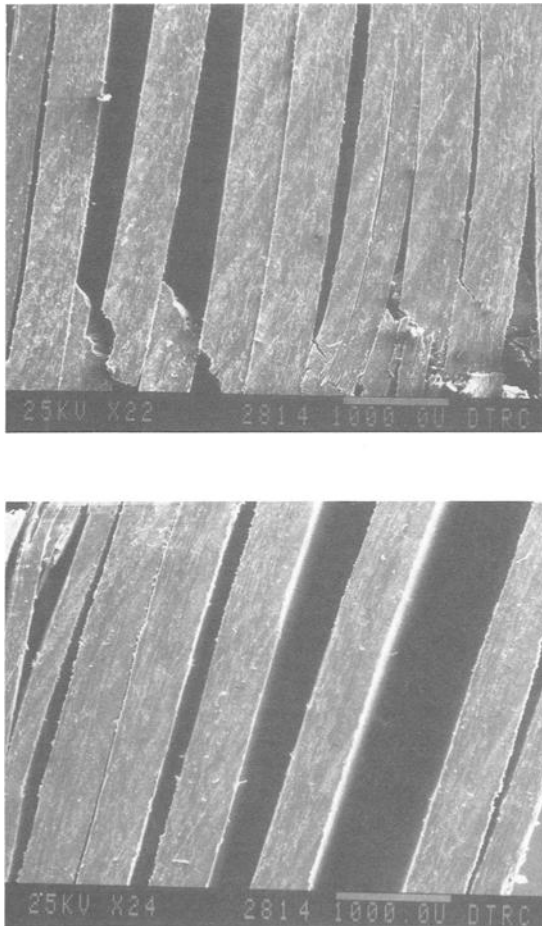


FIG. 10—Micrographs showing delamination patterns of outer plies in static and dynamic compressive specimens: (top) AC-2, static test ($\times 22$). (bottom) AC-8, dynamic test ($\times 24$).

Summary

From the limited number of specimens tested, it was found that reproducible strength and failure data were attainable using the described dynamic compression test method. Also, use of the endcap grips successfully prevented end-initiated brooming failure. Although “true” compression material property data for composites have been shown to vary depending on the test method and specimen geometry, the static data obtained by this test method are comparable to those obtained by other thick-section test methods. Because the fixture can be used for determining compressive strength properties for varied specimen sizes over a significant range of thickness and length, this method is useful as a comparative measure of high strain rate compressive properties for a variety of fiber-reinforced composite materials and laminate layups. Other dynamic methods, by contrast, are limited by inherent constraints in specimen size. Additionally, because the end constraint and specimen geometry for these high strain rate

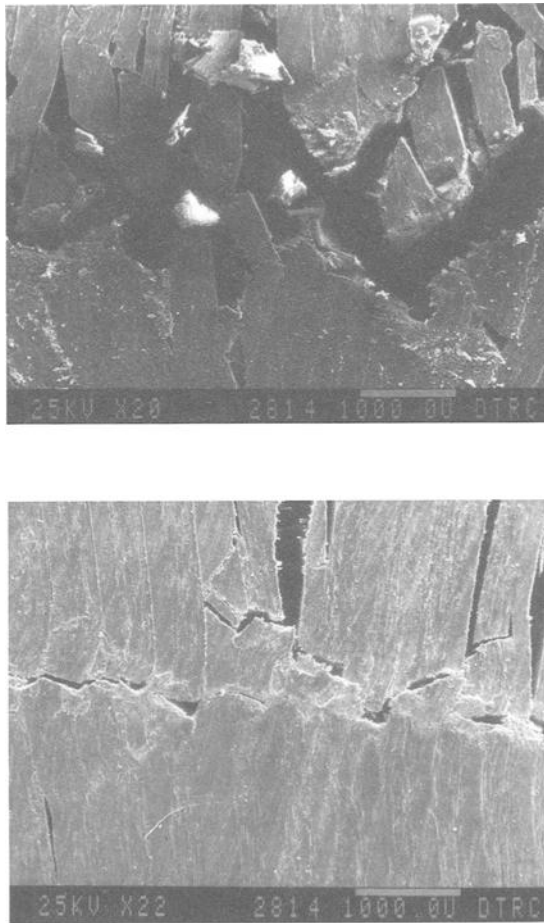


FIG. 11—Micrographs of crush zone regions in static and dynamic compression specimens: (top) AC-2, static specimen crush zone ($\times 20$). (bottom) AC-8, dynamic specimen crush zone ($\times 22$).

tests are the same as those employed in the static tests, this method can be used to compare directly the static and dynamic responses of a given material, allowing assessment of the effects of strain rate on that laminate system. Because of these capabilities, this high strain rate test method can provide a more realistic assessment of composite response to dynamic loads than other dynamic test methods.

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Session II

General Test Methodology Analysis and Development

Michael G. Castelli, Chairman

Overview

The development and analysis of new and innovative test methodologies for composite materials play a key role in our ability to use intelligently and confidently these materials in various structural applications. Defining a test methodology includes the development of testing techniques, such as specimen gripping, specimen design, displacement measurement, and specimen heating to name a few. When properly integrated into a test system and combined with appropriate experimental procedures, meaningful data can be produced for either development or verification. The manuscripts included from this session reflect some of the latest advances in metallic and polymeric composite testing.

“Thermomechanical Testing of High-Temperature Composites: Thermomechanical Fatigue (TMF) Behavior of SiC(SCS-6)/Ti-15-3” by Castelli, Bartolotta, and Ellis addresses the experimental difficulties involved with full thermomechanical testing. Various methods of specimen heating and specimen preparation are investigated to examine their influence on isothermal fatigue life. TMF conditions are shown to reduce drastically the high-temperature fatigue life of this titanium-based metal matrix composite. Damage and failure modes are also shown to be dependent upon the phase relationship between temperature and mechanical load.

“In-Plane Biaxial Compressive Deformation and Failure of E-Glass/Epoxy Laminates” by Doong, Faoro, and Socie introduces a unique test system for in-plane compression testing. Critical issues such as specimen-load alignment, specimen edge reinforcement, and lateral support conditions are discussed in detail. Calculated buckling stresses are compared to experimental results in unidirectional and cross-ply E-glass/epoxy laminates. Compressive strengths obtained from this biaxial test method are found to be much lower than those determined from the commonly used uniaxial IITRI method. Also, the specimens’ compressive strengths were found to be insensitive to the major/minor axis stress ratio.

“The Effect of Tab Orientation on the Distribution of Strains in Composite Specimens” by Foos, Wolfe, and Sandhu presents an experimental technique for minimizing non-uniformities in the stress/strain field of graphite/epoxy tabbed coupon specimens. Large non-uniformities can lead to premature failure outside the gage section and cast doubt on obtained tensile properties. A finite element analysis supported by experimental verification reveals that the strain field uniformity for any given laminate is dependent upon the angle that the tab line forms with the specimen. Tab angles are optimized on the basis of strain field uniformity for several laminates. Results show an increase in experimentally obtained tensile strengths and a reduction in the occurrences of tab region failures.

Thus, thermomechanical fatigue and biaxial compression testing and the effect of tab orientation on the test specimen are the general test methodology topics covered in this session that are important to the testing community.

Michael G. Castelli,¹ Paul Bartolotta,² and John R. Ellis²

Thermomechanical Testing of High-Temperature Composites: Thermomechanical Fatigue (TMF) Behavior of SiC(SCS-6)/Ti-15-3

REFERENCE: Castelli, M. G., Bartolotta, P. A., and Ellis, J. R., "Thermomechanical Testing of High-Temperature Composites: Thermomechanical Fatigue (TMF) Behavior of SiC (SCS-6)/Ti-15-3," *Composite Materials: Testing and Design (Tenth Volume)*, ASTM STP 1120, Glenn C. Grimes, Ed., American Society for Testing and Materials, Philadelphia, 1992, pp. 70–86.

ABSTRACT: Thermomechanical testing techniques recently developed for monolithic structural alloys were successfully extended to continuous fiber-reinforced composite materials in plate form. The success of this adaptation was verified on a model metal-matrix composite (MMC) material, namely SiC(SCS-6)/Ti-15V-3Cr-3Al-3Sn. The effects of heating system type and specimen preparation are also addressed. Cyclic lives determined under full thermomechanical conditions were shown to be significantly reduced from those obtained under comparable isothermal and in-phase bithermal conditions. Fractography and metallography studies of specimens subjected to isothermal, out-of-phase, and in-phase conditions revealed distinct differences in damage/failure modes. Isothermal metallography revealed extensive matrix cracking associated with fiber damage throughout the entire cross section of the specimen. Out-of-phase metallography revealed extensive matrix damage associated with minimal (if any) fiber cracking. However, the damage was located exclusively at surface and near-surface locations. In-phase conditions produced extensive fiber cracking throughout the entire cross section, associated with minimal (if any) matrix damage.

KEYWORDS: metal-matrix composites, thermomechanical testing, cyclic fatigue, SiC(SCS-6)/Ti-15-3, failure mode, fractography, heating effects, specimen preparation

Metal-matrix composites (MMC) and intermetallic-matrix composites (IMC) are being developed for a number of high-temperature aerospace applications. These composite materials offer superior strength/density ratios compared to monolithic materials, particularly at high temperatures. This level of performance typically is achieved by using high-strength ceramic fibers and low-density matrix materials. One obvious problem with this approach is that it combines materials with widely differing physical properties. This mismatch raises questions regarding the performance of MMC and IMC under the complex thermomechanical loading conditions typical of aerospace service.

The development of reliable thermomechanical testing techniques for this class of materials was viewed as a logical first step in addressing this concern. The approach adopted was to

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extend advanced testing techniques recently developed for monolithic structural alloys [1], such as Hastelloy Alloy X and Haynes 188, to composite materials in thin plate form. This effort included the development of computer test control software, an advanced heating fixture, a proper specimen design, and innovative experimental techniques. To verify the success of this adaptation, the full thermomechanical fatigue (TMF) behavior of a high-temperature composite was investigated.

Specialized real-time control software was developed to perform thermomechanical tests under load, strain, and combined load-strain control. One difficulty encountered with thermomechanical testing is ensuring precise phasing between the controlled variable, load or mechanical strain, and specimen temperature. This problem arises because of the response-time difference that typically exists between the temperature and the mechanical component of loading. Tests performed under conditions of improper phasing often produce distorted hysteresis loops, making subsequent data reduction and analysis difficult, if not impossible. The approach taken here was to incorporate subtle shifts in the *temperature-command* waveform at appropriate points in the cycle (that is, reversals). For example, the *temperature-command* is reversed prematurely such that the actual *temperature response* reverses simultaneously with the mechanical load. Through this “shifting” of the *temperature-command* waveform, the *temperature response* can be tailored and phased correctly with the mechanical component of loading.

Another difficulty encountered with uniaxial thermomechanical testing arises under strain control conditions. Under this control mode, thermal strains have a first-order effect on the mechanical loading component, namely, the axial strains. To control the mechanical strains properly, an accurate assessment of the thermal strains must be made at all times. This task is further complicated if phasing problems exist. The computer code extended for this study incorporates a function to calculate the real-time thermal strains based on a continuous analog-to-digital reading of the specimen's temperature. These calculated thermal strains are then appropriately added to the desired mechanical strains to obtain a total command strain. By this means, a nonlinear thermal expansion can be compensated for accurately to maintain the desired mechanical strain waveform.

Other important issues/problems include minimizing dynamic thermal gradients, temperature monitoring, and heating/cooling rates. Procedures were developed to optimize these interrelated variables with respect to specific equipment and test material. Adopting this approach, techniques originally developed for testing monolithic alloys have been extended successfully to composite materials.

Objective/Approach

The main objective of this work was to evaluate the extended thermomechanical testing techniques on a model MMC. A silicon carbide continuous fiber-reinforced titanium MMC designated SiC(SCS-6)/Ti-15V-3Cr-3Al-3Sn (Ti-15-3) was chosen as the material to be investigated. This system was selected because of availability, established processing parameters, and well-defined mechanical properties. Although the system is not typically thought of as a high-temperature material, it is useful as a “characteristic” material for developing and refining high-temperature testing techniques.

A second objective was to characterize the TMF behavior of SiC(SCS-6)/Ti-15-3. The room temperature [2,3], isothermal [4], and bithermal [5] behaviors of this composite system have been investigated. Also, Majumdar et al. [6] characterized the full TMF behavior of angle-ply laminates. However, very little (if any) information is available addressing the full TMF behavior of the unidirectional system.

There is some concern in the testing community about the effects of induction heating and

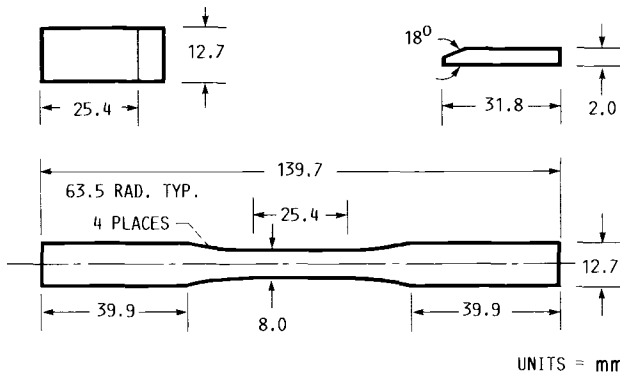


FIG. 1—Test specimen geometry and tab details.

method of specimen preparation on the fatigue life of the test specimens. These issues are briefly addressed through the comparison of several isothermal fatigue tests.

All material tested here is nine-ply laminate with 0° (longitudinal) fiber orientation. All test specimens had a nominal fiber volume ratio of 34% and a nominal thickness of 2 mm. The test specimens were cut from one 30.5- by 30.5-cm plate using wire electric discharge machining (EDM) followed by a light diamond grinding. Although the Ti-15-3 matrix material is a stabilized beta (bcc) phase, long-term exposures at elevated temperatures can lead to the precipitation of an α -Ti phase [7–9]. This primary hardening phase has been shown by several investigators to affect the mechanical properties. Therefore, to stabilize the material, the specimens were heat treated in a vacuum at 700°C for 24 h.³

The specimen geometry used for this investigation is shown in Fig. 1. The “dogbone” design features a 25.4-mm parallel section to facilitate close control of the temperature profile. Tabs were attached to the specimen with adhesive for assembly purposes and held in position by high clamping forces during the test.

All TMF experiments were conducted on a computer-controlled, servohydraulic test system; a close-up of the load train is shown in Fig. 2. Specimen heating was accomplished through direct induction work coils powered by a 5-kW radio-frequency heater. The induction coil fixture used here consists of three independently adjustable coil segments. This fixture, developed at NASA Lewis Research Center [10], provided the flexibility necessary to obtain closely controlled dynamic temperature profiles. Specimen cooling was accomplished through the combination of water-cooled grips and forced air. Although forced air creates difficulties when one is attempting to attain very precise temperature profiles, it was necessary here to obtain a target cycle period of 3 min. Intrinsic (spot-welded) K-type thermocouples were used for temperature monitoring and closed-loop control. The thermocouple locations did not exhibit premature cracking and, hence, had no influence on the fatigue life of the specimen. Axial strains were measured over a 12.7-mm gage section on the edge of the specimen by a high-temperature water-cooled extensometer. A selected number of baseline isothermal experiments were also conducted to investigate the effects of heating system type and method of specimen preparation. For these tests, a small two-zone direct resistance furnace was used to heat the specimens. All experiments performed in this study, as well as all data used for comparison, were conducted in an air environment.

³ Although the heat treatment improved the microstructural stability, it is believed that the material was still slightly metastable.

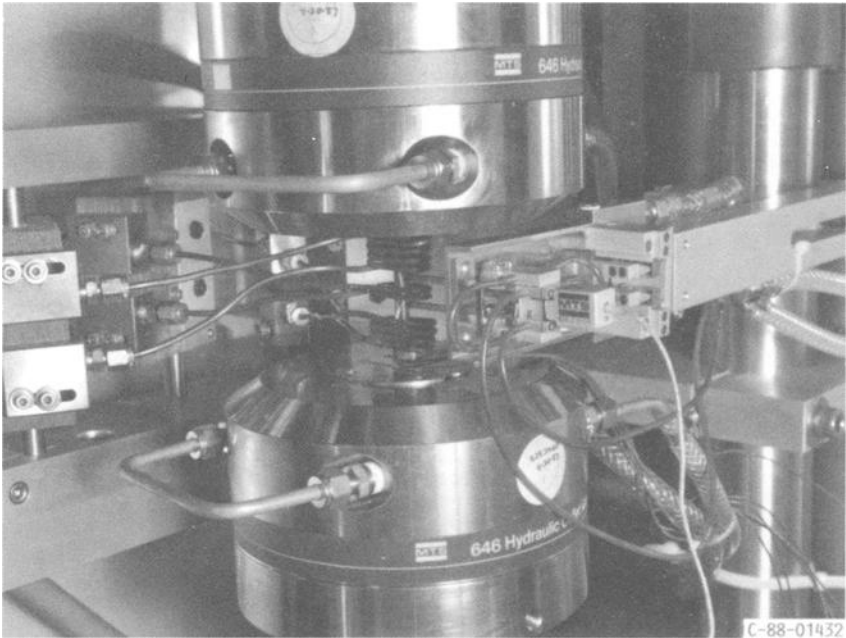


FIG. 2—Servo-hydraulic load train.

The test matrix performed for this study is given in Fig. 3. Specific stress levels were chosen to examine the low-cycle fatigue regime. All tests were run under load control. The isothermal tests were conducted at 427°C. A triangular load control function was cycled with a frequency of ten cycles per minute. As a result of the specimen’s geometry, compressive loads would induce buckling; therefore, all tests (including thermomechanical) were run to failure under

	MAXIMUM STRESS, MPa				
	1241	1034	965	896	827
ISOTHERMAL 427 °C (FURNACE HEATING)		1	1	1	
TMF OUT-OF-PHASE 93 °C - 538 °C (DIRECT INDUCTION HEATING)	2	2		2	2
TMF IN-PHASE 93 °C - 538 °C (DIRECT INDUCTION HEATING)		2	2	2	2

ALL TESTS WERE PERFORMED UNDER TENSION-TENSION CONDITIONS WITH A STRESS RATIO $R_0 = 0.05$.

FIG. 3—Test matrix.

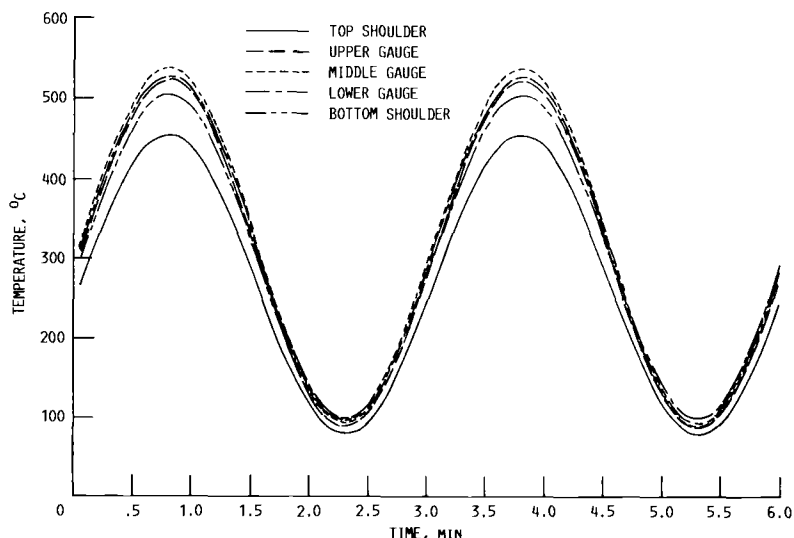


FIG. 4—Dynamic longitudinal temperature profile.

cyclic tension-tension conditions with a minimum/maximum stress ratio (R_σ) of 0.05. The TMF tests were conducted with both sinusoidal load and temperature control waveforms. Both in-phase and out-of-phase (180° phase shift between maximum load and maximum temperature) conditions were run with a cycle period of 3 min and a temperature range of 93 to 538°C.

As mentioned above, thermomechanical conditions introduced several new concerns associated with high-temperature testing. Previous testing performed on monolithic alloys [1] revealed that the dynamic thermal gradients were significantly greater than the static gradients existing before temperature cycling was initiated. Further, the dynamic gradient at any one location did not remain constant throughout the cycle, rather, it was found to fluctuate cyclically with the temperature magnitude. The degree of fluctuation was found to depend upon variables such as temperature range, temperature magnitude, and cycle period. To minimize dynamic thermal gradients, an appropriately imposed isothermal static gradient was established at some mid-cycle temperature. This technique is detailed in Ref 1. Figure 4 shows a typical dynamic temperature profile (longitudinal) taken from a TMF experiment. Note the variable dynamic gradient, illustrated by the changing distances between the respective curves. The three most closely grouped curves represent the gradients within the gage section, which experienced a maximum ($\approx 15^\circ\text{C}$) at the “hot” end of the cycle. This value was typical for all of the TMF tests.

The TMF test specimens were thermally cycled under zero load before mechanical loading to quantify the thermal strains and to ensure that a stable strain response existed. This cycling did not exceed ten cycles because the temperature-strain response was found to be initially stable.

Results and Discussion

Isothermal fatigue results are plotted as life curves on a stress basis in Fig. 5. Also shown in this plot are duplicate tests generated by Ervin et al.⁴ and similar tests generated by Gayda et

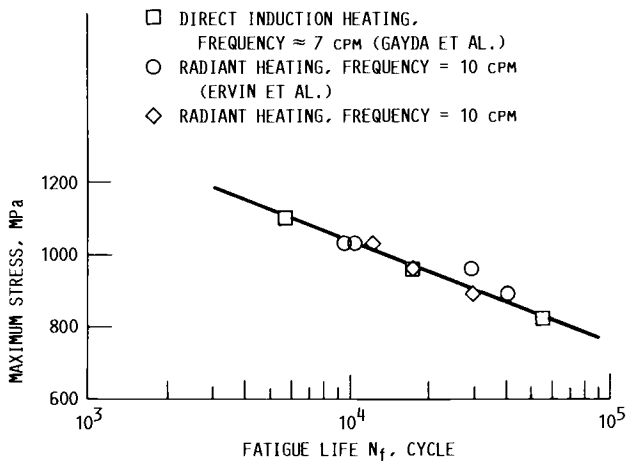


FIG. 5—Comparison of isothermal fatigue life of SiC/Ti-15-3 [0]₀, at 427°C.

al.⁵ The same material and specimen geometry were used for all the tests. In each case, failure was defined as complete fracture of the specimen into two pieces. The three sets of data compared were generated by a combination of specimen preparation and specimen heating methods. For specimen preparation, Gayda et al. used wire EDM cutting followed by light diamond grinding, whereas Ervin et al. used water jet cutting followed by hand polishing. For specimen heating, Gayda et al. used direct induction, and Ervin et al. used radiant heating. As shown in Fig. 5, the various cyclic fatigue lives are in very good agreement. This result was particularly surprising, given that the data were generated by three independent experimental groups.

An initial interpretation may be that the isothermal cyclic fatigue lives of this unidirectional composite material are not sensitive to the method of specimen preparation or method of heating. However, note that several issues of concern need to be addressed, particularly in regard to specimen design. The test specimens are cut from a large, continuously reinforced plate and, as a result, contain cut fibers along the edges. These locations are potential stress concentration/crack initiation points. Also, the specimens are tested with “as-received” surface conditions, which, from the traditional fatigue viewpoint, were less than ideal. Further, the majority of as-manufactured specimens exhibited curvature along their length, introduced into the original 30.5 by 30.5-cm plate during processing. Clearly, this curvature leads to nonuniform stress-strain states being developed in the specimens upon gripping and loading. Another area of concern is that the majority of the specimens failed at the blend between the parallel section and the beginning of the radius. Given these uncertainties, another interpretation of Fig. 5 may be that one or all of these effects play an important role in dictating the fatigue life of the test specimen. If this were the case, the plot would suggest that the effects promoted by method of specimen preparation and method of heating are secondary issues of concern, at least for the unidirectional material under investigation.

The stress-strain response for two of the isothermal tests is given in Fig. 6. Shown are the first and second cycle, and a cycle close to failure for tests performed with maximum stresses of 896 and 1034 MPa. The strain offset between these cycles naturally accumulated throughout the test. The cyclic strain ranges changed very little throughout the tests, suggesting minimal modulus degradation. However, as shown in Fig. 7, the cyclic mean strains significantly

⁵ J. G. Gayda et al., NASA Lewis Research Center, Cleveland, Ohio, Unpublished data obtained

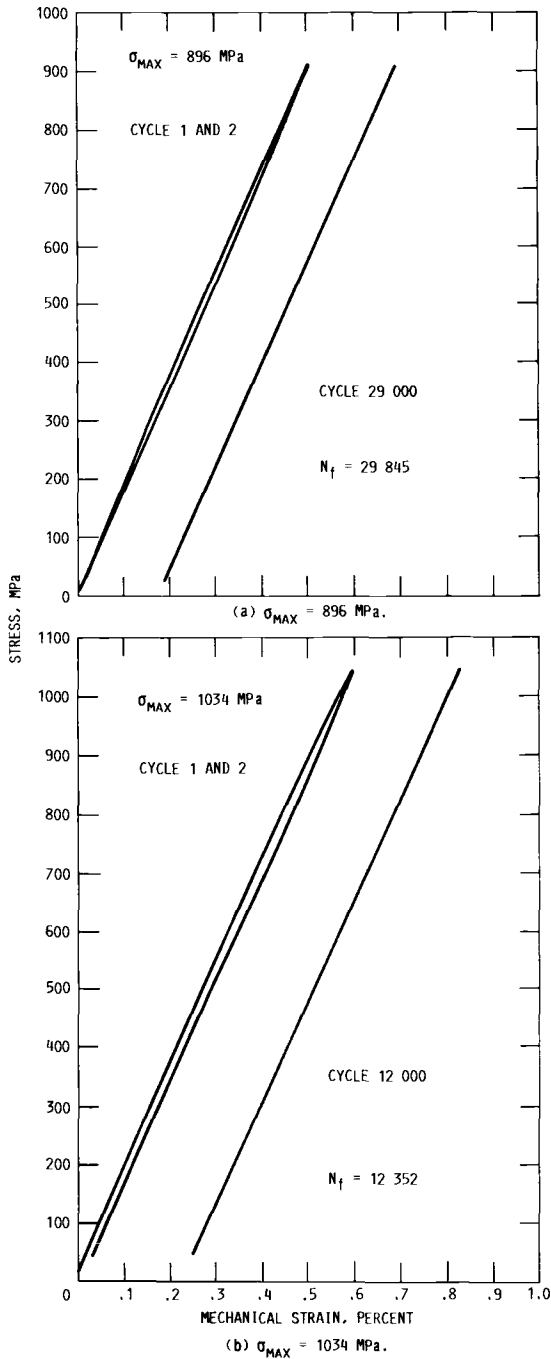


FIG. 6—Stress-strain response for two isothermal fatigue tests at 427°C: (a) $\sigma_{max} = 896 \text{ MPa}$ and (b) $\sigma_{max} = 1034 \text{ MPa}$.

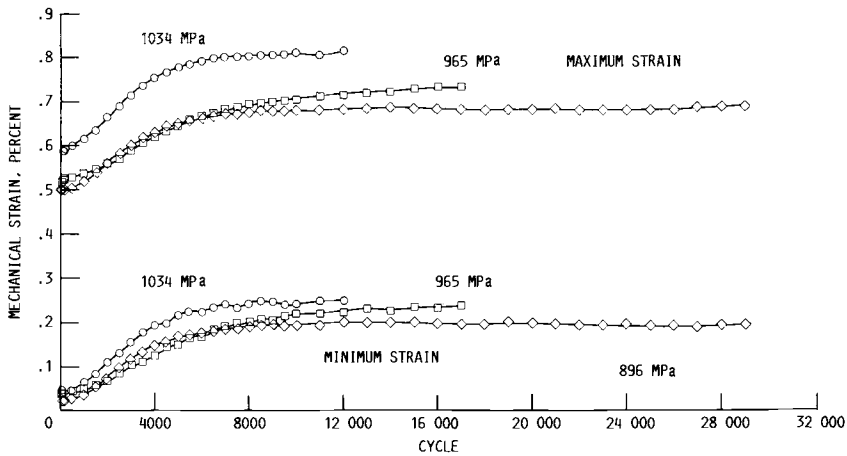


FIG. 7—Change in maximum and minimum strain during isothermal fatigue tests performed at 427°C.

increased over the first several thousand cycles and continued to increase gradually throughout the life of the specimens. This ratchetting effect was amplified as stress range was increased.

Typical scanning electron microscope (SEM) fractographs taken from the $\sigma_{\max} = 896$ MPa isothermal specimen are shown in Fig. 8. In general, the fracture surfaces from all of the isothermal specimens were very similar, revealing flat regions of fatigue cracking accompanied by randomly distributed areas of fiber pull-out and ductile matrix failure. The higher stressed specimens displayed a slightly increased amount of fiber pull-out and ductile matrix failure. Oxidation patterns and SEM backscatter imagery strongly suggest that matrix cracking was initiated at both surface and internal fiber/matrix interface locations. Matrix “river” patterns suggest that there were numerous crack initiation points and that the cracks have propagated only short distances ($\approx$ one to two fiber rows) before coalescing. This relatively short crack growth may be a result of the fibers hindering fatigue crack growth.

Results from the TMF tests are plotted in Fig. 9 as life curves on a stress basis. For reference purposes, the isothermal fatigue line displayed in Fig. 5 is also shown in Fig. 9. In general, cyclic lives determined under full TMF conditions were greatly reduced from those obtained under comparable isothermal conditions. Also, for the case of high-stress, in-phase loadings, lives were reduced by several orders of magnitude from those obtained under comparable isothermal and bithermal (non-isothermal) conditions. This magnitude of degradation is best illustrated in Fig. 9 by a line drawn through $\sigma_{\max} = 1034$ MPa. The 538°C isothermal datum point was obtained from a specimen subjected to fatigue cycling with a 3-min cycle. This condition was imposed to address the effects (increased ratchetting, environmental) of the relatively slow cycle period. The isothermal specimen experienced a fatigue life of approximately 4000 cycles, whereas the specimens subjected to true in-phase loading conditions from 93 to 538°C consistently failed on the second load-up.

The large magnitude of life degradation resulting from true in-phase conditions was not seen under in-phase bithermal (non-isothermal) conditions [6] where a three- to fourfold decrease from isothermal life was obtained. The decreased life displayed under full TMF conditions may be attributed to factors (not present under bithermal conditions) promoting minimized reversed inelastic strains in the matrix and thus increased ratchetting (discussed later). One obvious condition promoting damage is that attributed to the coefficient of thermal expansion (CTE) mismatch. Work performed by Gabb et al. [5] on this MMC system revealed that spec-

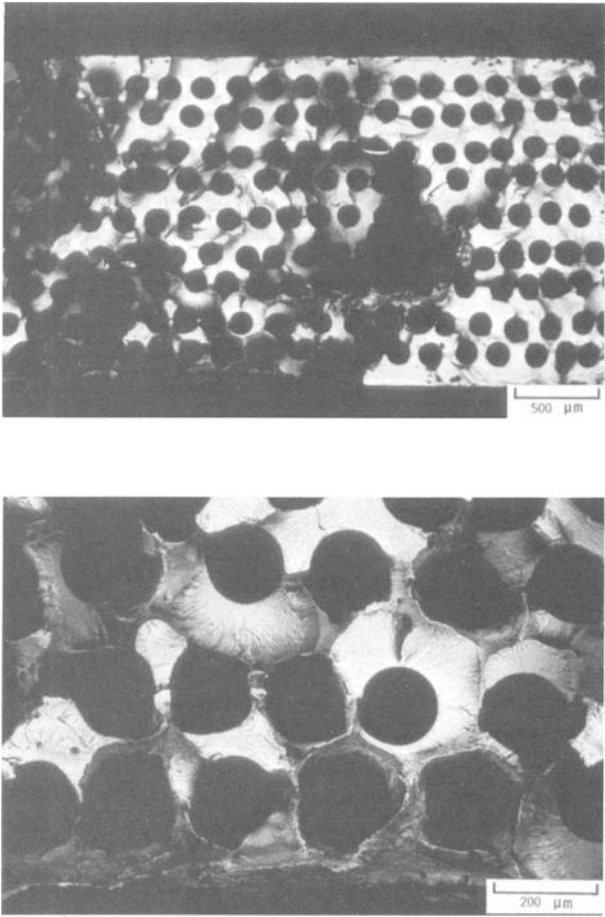


FIG. 8—Typical SEM backscatter fractographs taken from the $\sigma_{max} = 896$ MPa isothermal test performed at 427°C: (a) $\times 30$ and (b) $\times 100$.

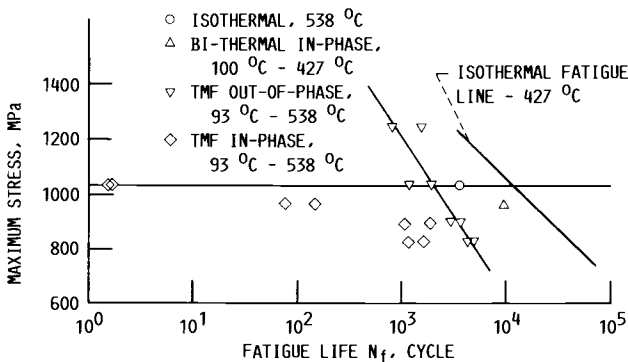


FIG. 9—Thermomechanical fatigue life results for SiC/Ti-15-3, [0].

imens subjected to 10 000 thermal cycles (from 300 to 550°C) under zero load did not exhibit degraded mechanical responses during subsequent tests. However, the CTE mismatch did promote highly localized cracks in the reaction zone and carbon-rich coating of the fibers. True thermomechanical conditions introduce this damaging effect from the CTE mismatch as a dynamic component compounded with the transverse Poisson's effects associated with mechanical loading.

The stress-mechanical strain response for an out-of-phase and in-phase test with $\sigma_{\max} = 896$ MPa is shown in Fig. 10. Note that the hysteresis loops do not exhibit irregular distortions. This is an essential check in verifying that the thermomechanical test is well controlled. Again, the first two cycles and a cycle just before failure are plotted, where the mechanical strain offset is that which naturally occurred throughout the test.

In general, the out-of-phase hysteresis loops (Fig. 10a) were open, displaying evidence of inelastic deformation in any given cycle. Of the three conditions (isothermal, out-of-phase, and in-phase), only the specimens subjected to out-of-phase loadings displayed a noticeable change in compliance. This change in compliance is evidenced by an overall change in slope of the TMF loop. Out-of-phase TMF involves "cold" temperatures at high stress and "hot" temperatures at low stress. This condition is not conducive to creep-ratchetting effects, and thus, as shown in Fig. 11, the behavior reveals gradual increases in mean strain at the onset of the test. This increase is followed by a subtle decrease in mean strain as a possible result of matrix work hardening or the precipitation of the α -Ti phase or both. Also, specimens subjected to out-of-phase conditions displayed substantial increases in mean strain late in life ($N > 0.7 N_f$), suggesting failure was approaching.

In-phase thermomechanical cycling involves "hot" temperatures at high stress and "cold" temperatures at low stress. This, combined with the specific test parameters selected for the TMF experiments, results in conditions that are highly conducive to creep ratchetting. Specifically, several parameters can be identified: (1) high stress coinciding with high temperatures, (2) a load control environment where strain limitations are not imposed, (3) a sinusoidal control waveform that lingers at the function's extremes, (4) a relatively slow cycle period of 3 min, and (5) a tension-tension cycle in which reversed inelastic strains are minimal. The ratchetting attributed to minimized reversed inelastic strains is further amplified by an increasing elastic modulus (as a result of a temperature decrease) during unload. This "cold," "nearly elastic" unload is absent under in-phase bithermal conditions [6] and, hence, may contribute to the differences in cyclic fatigue lives discussed earlier.

As shown in Figs. 10b and 12, the in-phase loading response exhibits an extensive increase in mean strain (creep ratchetting) during the initial cycles. This "unrestrained" ratchetting is believed to be the primary cause of the dramatic decrease in fatigue life when compared with out-of-phase TMF conditions. At 538°C, the titanium-based matrix creeps readily [11]. At these high temperatures, the matrix is unable to carry its share (on a rule of mixtures basis) of the load, therefore, it relaxes as if in a strain-controlled environment. The load shed by the matrix is distributed to the fibers, causing an increased fiber stress and composite mean strain. This process gradually continues at a decreasing rate (as a result of hardening of the matrix) until eventual fiber overload occurs. This process occurs under isothermal [5] and bithermal [6] conditions but is greatly accelerated under in-phase TMF conditions as a result of the factors leading to minimal reversed inelastic strains. The mean strain response stabilized early in life and did not experience an increase late in life as seen under out-of-phase conditions (compare with Fig. 11); thus, there was no phenomenological indication that failure was approaching. Also, as evidenced in Fig. 10b, there was no detectable change in compliance for specimens subjected to in-phase loadings.

Typical SEM fractographs taken from a specimen subjected to out-of-phase loading con-

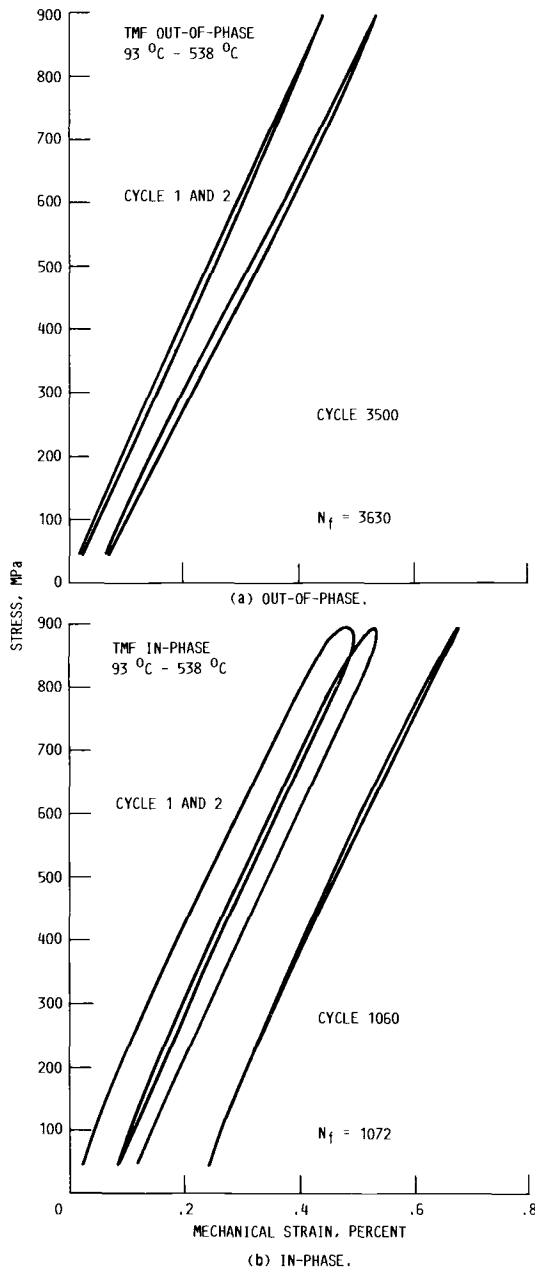


FIG. 10—Stress-mechanical strain response for two thermomechanical tests performed at $\sigma_{max} = 896$ MPa: (a) out-of-phase and (b) in-phase.

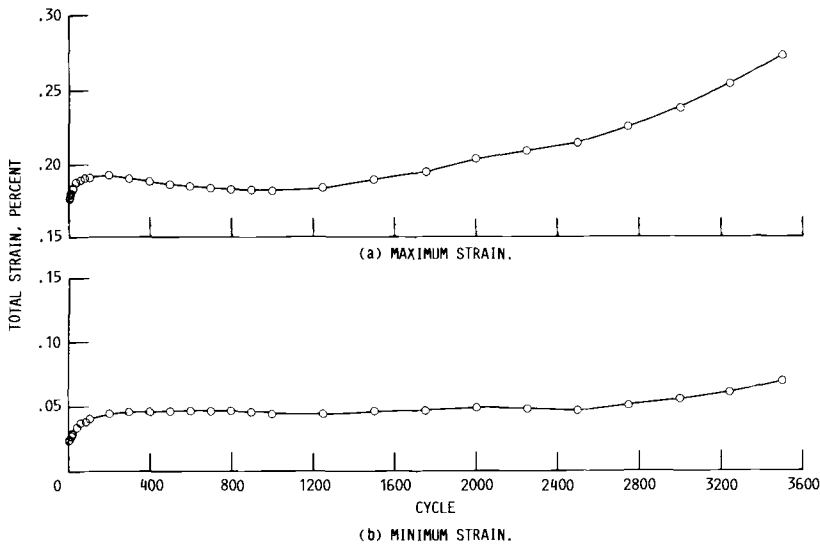


FIG. 11—Change in maximum and minimum strain for an out-of-phase test with $\sigma_{max} = 896$ MPa.

ditions, $\sigma_{max} = 896$ MPa, are shown in Fig. 13. The fracture surface reveals a fair amount of fatigue cracking, but, in general, the area associated with tensile overload appears to be larger than that found in comparable isothermal specimens. True thermomechanical conditions appear to have introduced crack initiation and propagation mechanisms different from those found under isothermal conditions. Oxidation patterns and SEM imagery clearly suggest that crack initiation sites occurred almost exclusively at surface and near-surface fiber-matrix interface locations. Out-of-phase conditions consistently produce fracture surfaces displaying an

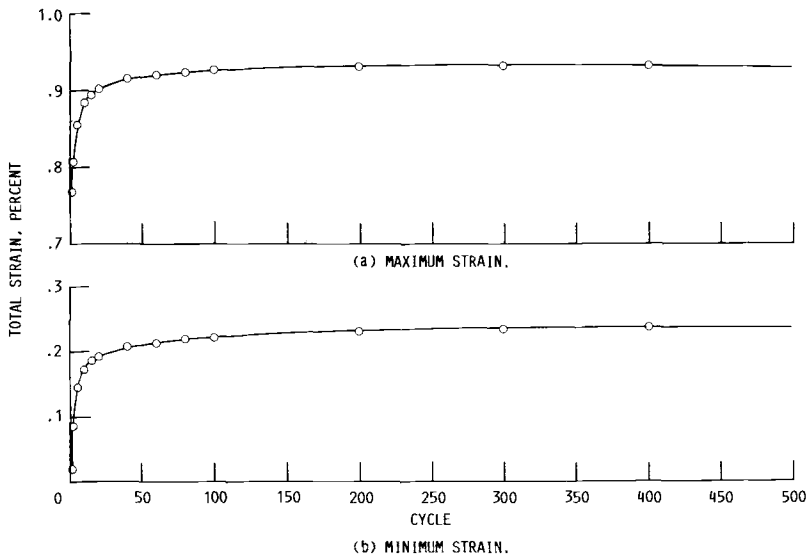


FIG. 12—Change in maximum and minimum strain for an in-phase test with $\sigma_{max} = 896$ MPa.

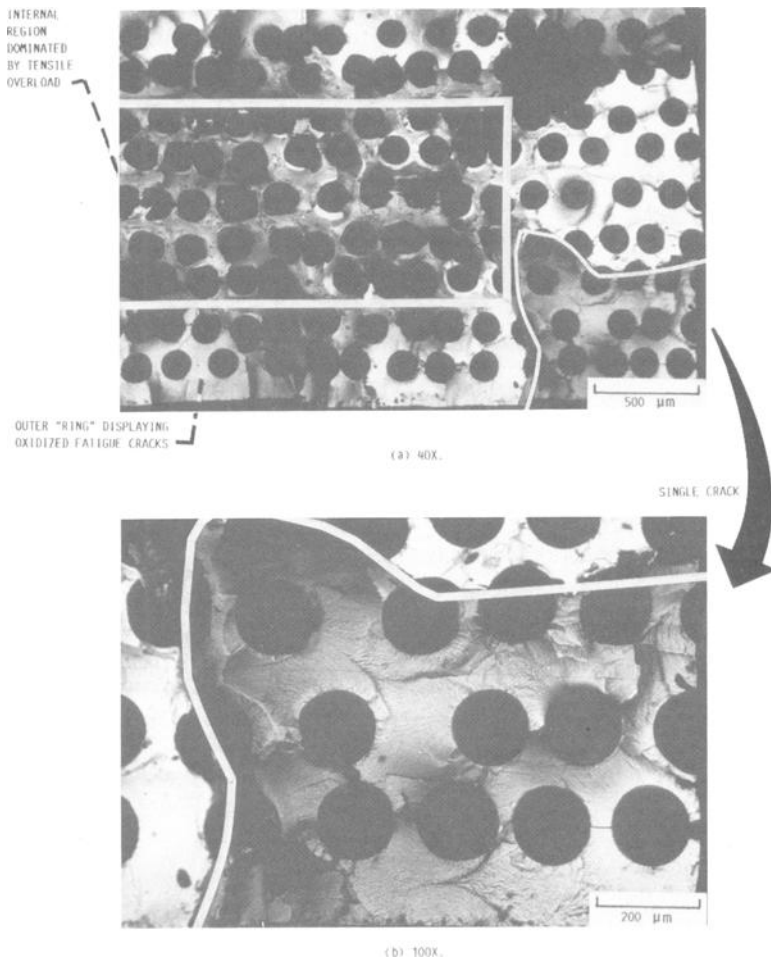


FIG. 13—Typical SEM backscatter fractographs taken from a specimen subjected to out-of-phase loading with $\sigma_{\max} = 896 \text{ MPa}$: (a) $\times 40$ and (b) $\times 100$.

outer “ring” of oxidized cracking (approximately two to three fibers into the cross section) surrounding an internal region exhibiting ductile matrix failure and fiber pull-out. This suggests that the surface plays a decisive role in the damage processes. Also, matrix “river” patterns reveal that individual cracks are less numerous (compared to isothermal fracture surfaces) and appear to have propagated greater distances, as illustrated upon comparing Figs. 8 and 13. Under out-of-phase conditions, the fibers appear less effective at arresting the crack growth.

Typical longitudinal and transverse sections taken from an out-of-phase specimen ($\sigma_{\max} = 896 \text{ MPa}$) are shown in Fig. 14. The sections reveal that the primary cyclic fatigue damage consists of transverse matrix cracking. This cracking initiated at surface locations and near-surface fiber-matrix interfaces. Very little fiber damage was visible. A matrix-dominated failure initiated by matrix microcracking at these locations, is compatible with the fact that the

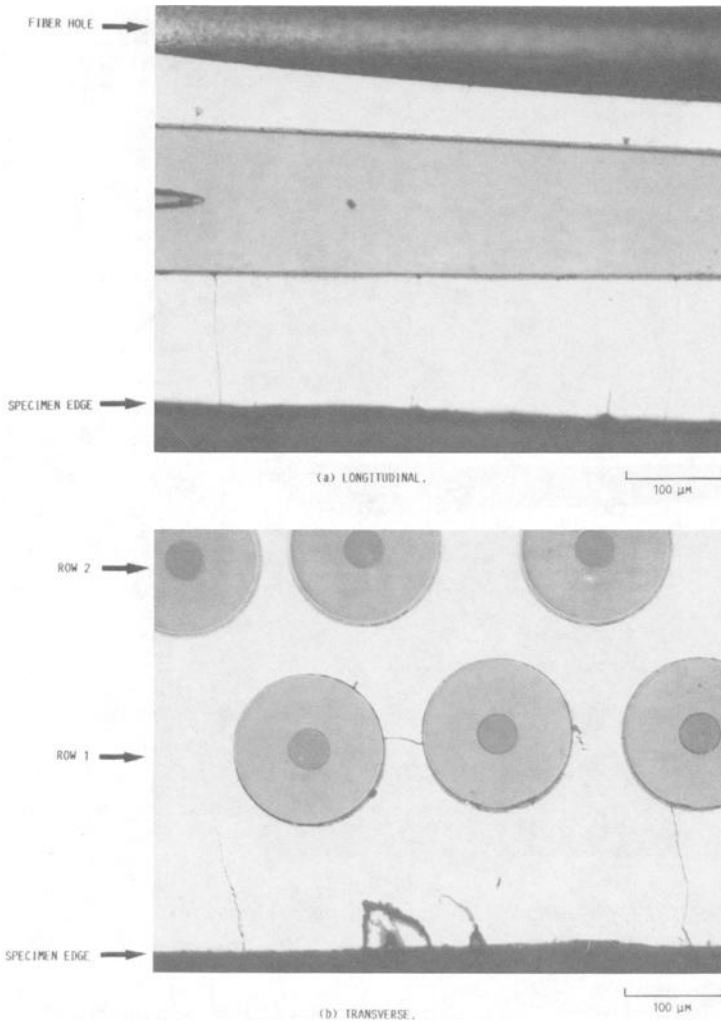


FIG. 14—Typical longitudinal (a) and transverse (b) sections taken from a specimen subjected to out-of-phase loading with $\sigma_{\max} = 896$ MPa.

out-of-phase conditions maximize the longitudinal stresses in the matrix; matrix stresses produced by the mechanical loading and thermal expansion mismatch are additive.

In contrast, metallography from specimens subjected to in-phase TMF conditions exhibits trends quite different from those seen in isothermal and out-of-phase TMF specimens. Typical SEM fractographs taken from an in-phase specimen, $\sigma_{\max} = 896$ MPa, are shown in Fig. 15. Figure 15 reveals extensive fiber pull-out and ductile matrix failure across the entire fracture surface. In fact, the features illustrated here (and from all other in-phase tests performed) most closely resemble those displayed by specimens subjected to tension tests [12]. This pattern of pure tensile overload and lack of matrix cracking was a somewhat surprising result, given that several of the (lower stress tests) experienced well over 1000 cycles.

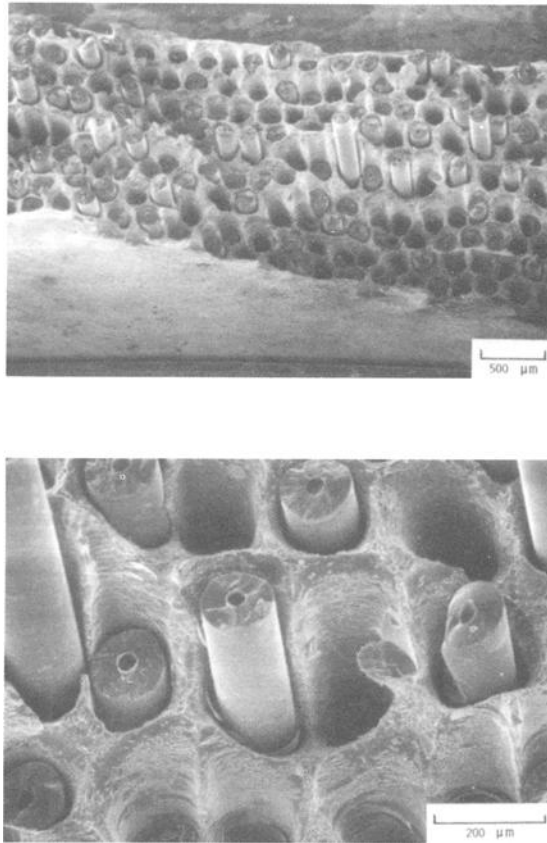


FIG. 15—Typical SEM fractographs taken from a specimen subjected to in-phase loading with $\sigma_{max} = 896 \text{ MPa}$: (a) $\times 30$ and (b) $\times 100$.

Longitudinal, shown in Fig. 16, and transverse sections taken from the gauge section (away from the fracture surface) clearly reveal extensive fiber damage in the absence of matrix cracking (a scenario completely opposite to that displayed by the out-of-phase metallography). Fiber cracking was found in both the transverse (more predominant) and longitudinal directions. The fiber-dominated failure found under in-phase conditions is also compatible with the fact that in-phase conditions maximize the longitudinal stresses in the fiber, whereas fiber stresses produced by mechanical loading and thermal expansion mismatch are additive.

Because the results presented here are influenced by the particular test parameters, note that dramatically different material behavior and resulting fatigue lives can be produced by changing these test parameters. By changing the control mode from load to strain, several of the prime factors leading to extreme creep ratchetting during in-phase TMF can be eliminated. A similar argument can be used when considering the use of a triangular control function, rather than the sinusoidal used here, because the sinusoidal function lingers at the extremes. Further, the advent of computer-controlled testing has enabled any combination of mixed-mode and mixed-function control. For example, a load-controlled test exhibiting a fixed stress rate can be performed with limiting strain boundaries. The ongoing debate addressing the appropri-

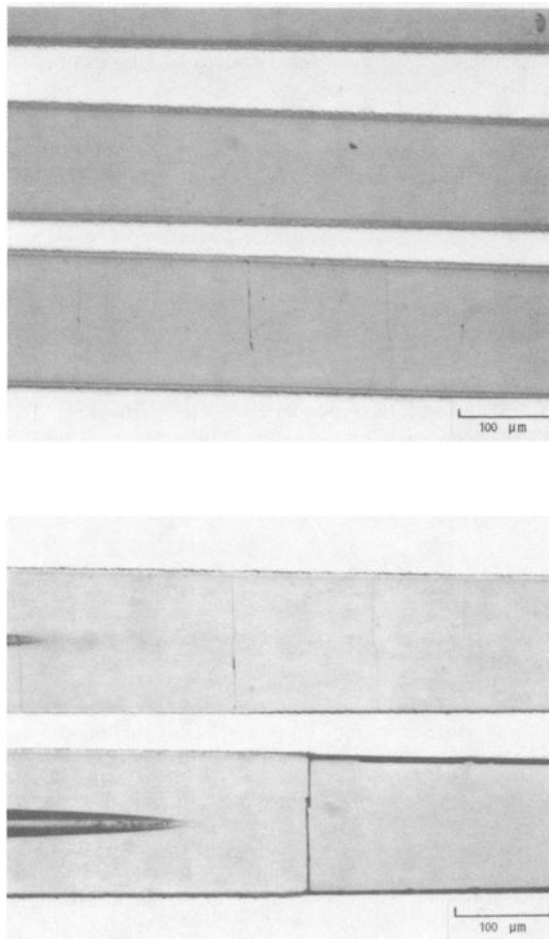


FIG. 16—Typical longitudinal sections taken from a specimen subjected to in-phase loading with $\sigma_{max} = 896 \text{ MPa}$: (a) through the width and (b) through the thickness.

ateness of “load control” or “strain control” will most likely never be resolved. However, “test-induced” effects, such as creep ratchetting, should be an issue of concern, both before the tests are performed and during subsequent data interpretation and analysis.

Conclusions

Improved thermomechanical testing techniques developed for monolithic materials were successfully extended to SiC/Ti-15-3 in plate form. As a result, closely controlled TMF tests were performed with minimal experimental uncertainties. Isothermal fatigue life data did not reveal any sensitivity to heating method (direct induction versus radiant) or specimen preparation method (water cutting/hand polishing versus wire EDM/light diamond grinding). This was evidenced by the close agreement existing among tests generated by three independent experimental groups. However, several key concerns were identified as potentially having a

degrading effect on the cyclic fatigue life of the test specimen. These issues are currently being addressed through the investigation of new specimen geometries that minimize stress concentrations, as well as through techniques that lead to test specimens containing completely encapsulated, uncut fibers.

The isothermal data exhibited an essentially linear response after the first cycle with no evidence of modulus degradation throughout the test. However, significant increases in mean strain developed, with the majority of this occurring early in the cyclic life. Metallography revealed a large amount of flat fatigue cracking throughout the cross section of the specimen.

Cyclic fatigue lives were greatly reduced under TMF conditions when compared to isothermal and in-phase bithermal data. This was particularly the case for in-phase loadings at high stresses where lives were decreased by several orders of magnitude.

Failure modes displayed under TMF conditions were significantly different from those displayed under comparable isothermal conditions. Out-of-phase loadings clearly displayed a matrix-dominated failure, whereas cracking and crack initiation points were found exclusively at surface locations and near-surface interface locations. In-phase TMF loadings displayed fiber-dominated failures, whereas extensive fiber damage was found in the absence of matrix cracking.

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In-Plane Biaxial Compressive Deformation and Failure of E-Glass/Epoxy Laminates

REFERENCE: Doong, S-H, Faoro, J E, and Socie, D F, "In-Plane Biaxial Compressive Deformation and Failure of E-Glass/Epoxy Laminates," *Composite Materials Testing and Design (Tenth Volume)*, ASTM STP 1120, Glenn C. Grimes, Ed., American Society for Testing and Materials, Philadelphia, 1992, pp. 87-102

ABSTRACT: This paper presents the development of a test system and associated experimental procedure for evaluating in-plane biaxial compressive deformation and failure of fiber composite laminates. A description of the test system and the design of fixtures and specimen are given. Critical buckling stress of the specimen is determined analytically and compared to the experimentally obtained compressive strength. The capability of this testing system and accuracy of the experimental results are discussed and evaluated. A series of tests have been conducted on E-glass/epoxy composite laminates with different lamination variables and stress biaxiality ratios. Compressive failure strengths of composite materials are determined in various in-plane biaxial loading conditions.

KEYWORDS: composites, laminates, multiaxial testing, compressive strength, deformation, failure

In recent years, advanced fiber-reinforced, polymeric-matrix composite laminates have been considered as candidate materials for constructing large submersible structures. The loading conditions of these structures require a comprehensive database for composite materials under in-plane biaxial stresses. Reliable design of the composite structures clearly requires a thorough understanding of failure mechanisms in composite laminates under complex biaxial stress states. Because no experimental results are available in the open literature for investigation of biaxial compressive failure behavior of fiber composite materials, development of a test system capable of applying general biaxial compressive stresses to composite laminates is a first and key step to achieve these goals.

A number of difficulties have been identified in previous efforts to conduct biaxial testing on planar composite specimens [1-3]. Bert et al. [1] used a flat cruciform specimen to determine the failure strengths of glass fiber woven cloth-reinforced epoxy laminates under biaxial tension. Although a uniform stress field was achieved in the gage section of the specimens, biaxial stress values were determined from measured strains. It was unclear how material nonlinearity might affect accuracy of the strength values. Duggan [2] introduced slots in tension-compression specimens to create a statically determinate stress field. With these slots, major failure could initiate from high stress concentration areas. The effects of this stress concentration may vary with the stress biaxiality ratios and change qualitative features of the experimentally determined failure envelopes. Designing specimens that consist of a statically deter-

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minate stress field without stress concentration sites is a great challenge to the biaxial testing of planar composite laminates.

This paper presents the experimental details of an in-plane biaxial compressive testing system developed at the University of Illinois at Urbana-Champaign. In this paper, the experimental setup of the testing system is described first. This is followed by experimental results on E-glass/epoxy composite laminates. Based on these results, the stress homogeneity, buckling condition, and proportionality of the loading histories are investigated. This study shows how stringent requirements of planar biaxial specimen can possibly be satisfied for in-plane biaxial compression tests.

Experimental Procedure

Testing System

The in-plane biaxial testing system consisted of a square frame (Fig. 1), which had one linear actuator located on each side of the frame. Each actuator was capable of applying a maximum of 450 kN of force to the specimen, which was at the center of the testing frame. Motion of the linear actuators could be controlled either by displacement of the actuator or by the amount of force detected by the load cell. The actuators at two opposite sides of the testing frame were designed to have the same amount of displacement or force. The displacement, or force, of the actuators along one loading axis was independent of the displacement, or force, along the other axis. This system was controlled by a microcomputer for testing and data acquisition.

In this study, all experiments were conducted in displacement-control so that the fracture surface of the specimen could be preserved for examination. Changes in stress biaxiality were

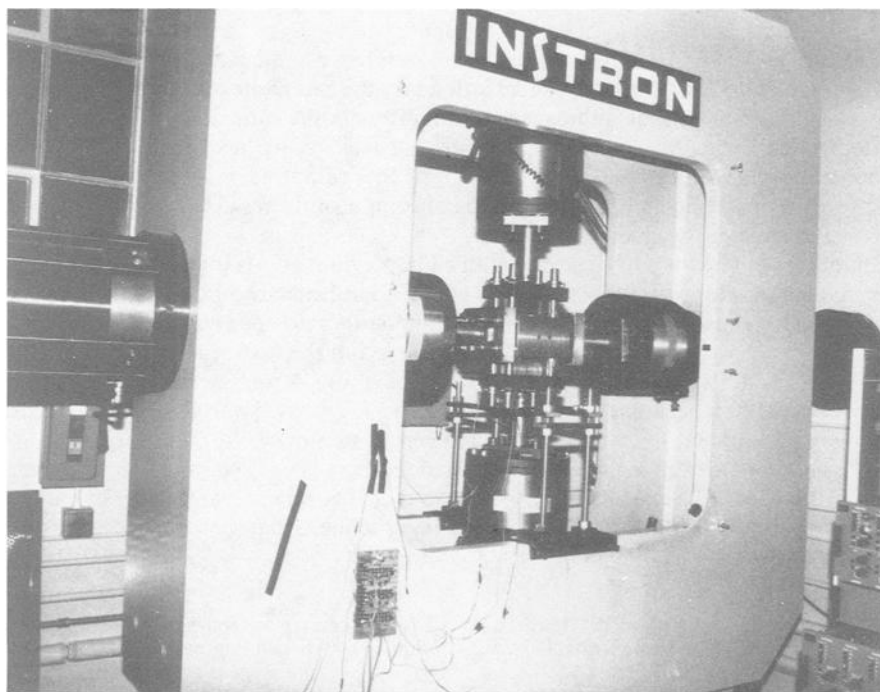


FIG. 1—In-plane biaxial testing system with compressive testing fixture.

achieved by changing ratio of the displacements of linear actuators along the two axes. The displacement rate of these actuators ranged from 0.03 to 0.05 mm/s. Because the stress increment along one axis depended on the directional stiffness of the specimen and specimen fixture, the stress increments were not proportional although displacements of the linear actuators were proportional. Thus, stress biaxiality changed throughout the loading history of one specimen.

Material

An E-glass fiber-reinforced epoxy resin with a tradename of Scotchply Reinforced Plastic® was used. This material had a fiber volume of 64% by weight. Composite laminates were made from the uncured unidirectional prepregs with two stacking sequences. The first one consisted of ten layers of prepregs aligned along the same direction, which is designated as $[0^\circ]_{10}$ in the following text. Another kind of laminate also contained ten plies of unidirectional prepregs. Six of these ten plies were, however, perpendicular to the direction of the other four plies. This laminate is called cross ply and designated as $[0^\circ/90^\circ/0^\circ/90^\circ/0^\circ]_2$. These composite laminates were made by pressing at 170°C for 30 min. Thickness of the laminates was 2.54 mm, and the variation was less than 0.1 mm.

Specimen Design

The biaxial specimen was a 127-by 127-mm square composite panel (Fig. 2). Biaxial compressive stress state was achieved by applying compressive loading at four sides of the specimen. Compressive forces were applied by direct contact at the specimen edges. To prevent edge failure, each edge of the specimen was reinforced with three metal plates. As shown in Fig. 3 two of these three metal plates, Plates 1 and 2, were attached to two sides of the specimen.

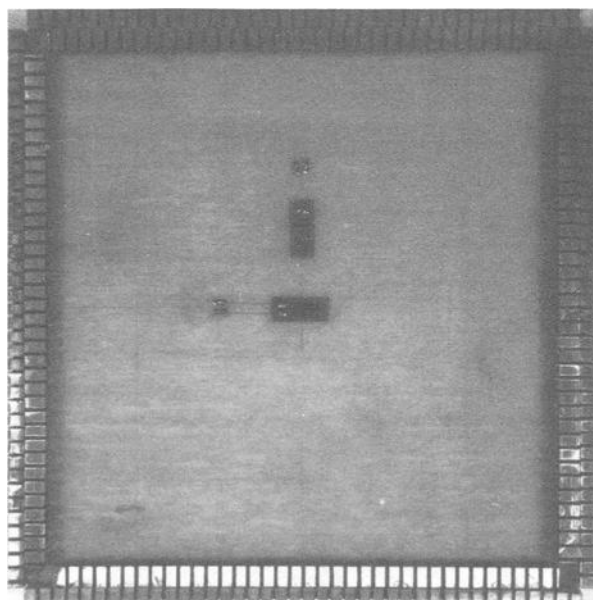


FIG. 2—In-plane compressive test specimen with flexible tabs.

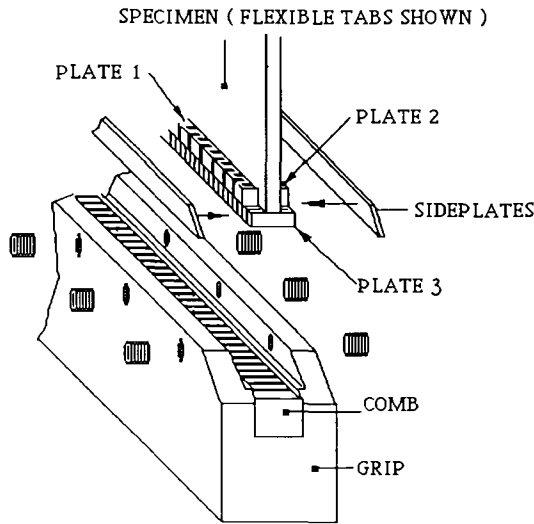


FIG 3—Edge reinforcements on specimen edges

By applying a small amount of clamping forces from these two plates, the specimen was protected from being crushed at edges. Metal Plate 3 was attached to the cross-sectioned area of the specimen. This plate served as a medium for transferring loading to the specimen. Because compressive forces were applied on four sides, the metal plates used to protect specimen edge from being crushed by loading along one axis, the x axis, could become load-bearing members for the compressive loading along the other axis, the y axis. As a result, the combination of specimen and these metal plates became a composite structure and the stress inside the specimen could not be directly determined from the applied loading. To overcome this problem, a large number of slots were cut in the metal plates (see Fig. 3). With these slots, the metal plates retained their strength in a direction normal to the edge of the specimen. However, stiffness of the metal plate became so low (10 kN/m) in the direction parallel to the edge of the specimen that all the compressive forces would be taken by the specimen itself. The stresses inside the specimen could then be calculated from the applied force directly. The adequacy of this method has to be examined with experimental results, which are addressed later.

Another reason for cutting many slots in the reinforcing plates was to ensure stress homogeneity at corners of the specimen. It is well known that high shear stress concentration occurs in areas close to edges of the compressed areas [4]. To avoid such stress concentration, compressive stress was designed to cover the entire specimen edge. With this design, the metal plates attached to adjacent sides of the specimen touch each other before compressive loading was applied. During compressive testing, flexibility of the slotted metal plates along the specimen edge allowed these plates to reduce their lengths so that the areas close to corners of the composite specimen could be compressed homogeneously. Corner failures were not observed during the testing.

Specimen Fixture

Alignment of the loading axes was a critical issue for the success of material testing with compressive loading. For a testing system such as the one used in these experiments, precise alignment of the loading axes could not be easily accomplished by adjusting locations of the linear actuators. For this reason, a specially designed specimen fixture (Fig. 4) was used. Com-

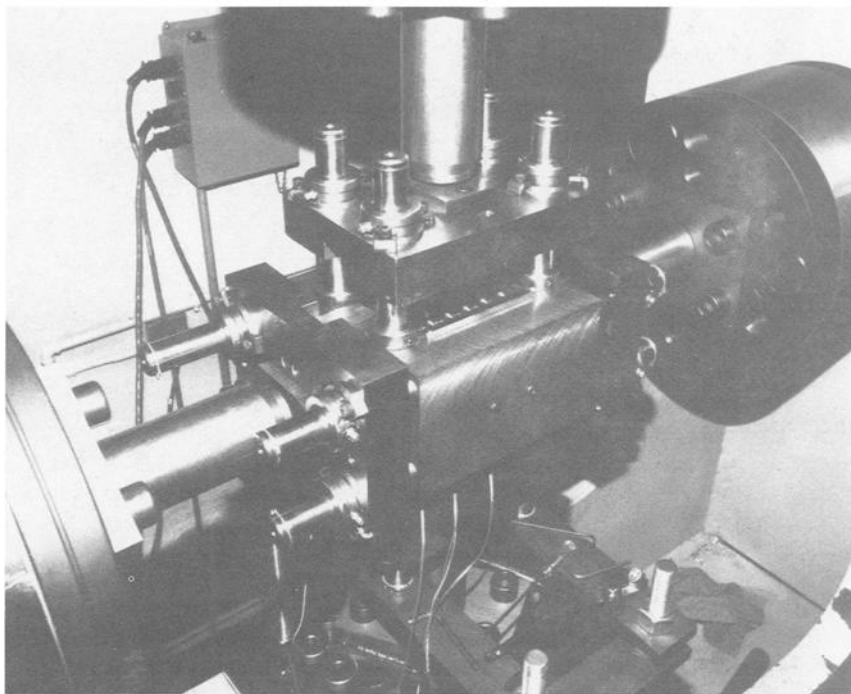


FIG. 4—Compressive test fixture.

pressive forces were transferred to the composite specimen through four steel blocks. Movement of these steel blocks was guided by eight round bars to form two perpendicularly oriented loading axes.

Because the purpose of these experiments was to determine material strength, specimens had to be protected from structural buckling before they reached the material failure strength. Two issues were considered important in designing the lateral supports for preventing specimen buckling. First, the friction stress between lateral supports and surface of the specimen had to be as low as possible. This was to avoid creating in-plane force by lateral supports, which might decrease accuracy of the calculated compressive stresses. At the same time, the lateral supports had to be stiff along the direction normal to the surface of the specimen to prevent lateral displacement. Second, the number of lateral supports had to be enough to modify the buckling mode of the specimen to a stage at which the critical buckling stress of the specimen was higher than the material failure strength. Although this consideration could be resolved by either reducing the specimen size or increasing the number of supports, the latter option was chosen to have a relatively large specimen. Specimen size was important for studying the size effects of composite materials.

With the aforementioned considerations, the lateral supports used in these experiments were 30-mm-long, 7-mm-diameter steel bolts with their heads being capped with Teflon® tabs (see Fig. 5). The diameter of these bolts was reduced on both ends to 3 mm to make them relatively flexible along the loading directions. Friction forces between the specimen surface and the lateral supports were small. The Teflon caps were used to prevent the specimen surface from being damaged by these lateral supports. Because thickness of the composite specimen could not be controlled precisely for most of the composite laminates, every support had to be adjusted independently. This was to ensure contact between each Teflon cap and the specimen surface.

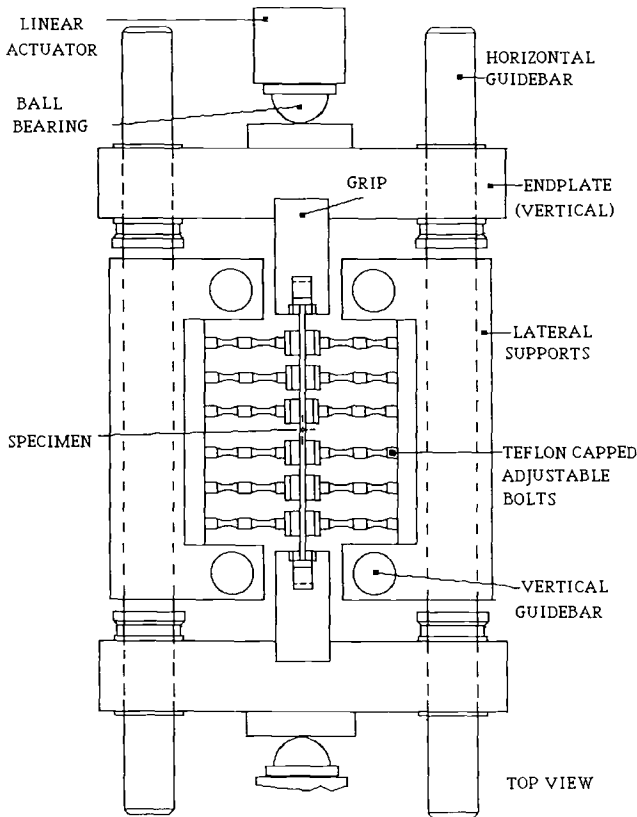


FIG 5—Top view of test fixture

In designing the number of lateral supports, a simplified buckling analysis was made. It was assumed that the critical buckling stress of the specimen could be bounded with three simple models. The first model was a square plate with its dimension equal to the distance between center lines of two adjacent supports. Four edges of this plate were assumed to be simply supported, and compressive stresses were applied from four directions with different stress ratios. The critical buckling stress for this case is [5]

$$N_1 = - \frac{\pi^2 [D_{11}m^2 + 2(D_{12} + 2D_{66})n^2 + D_{22}n^4/m^2]}{a^2 (1 + \alpha(n/m)^2)} \quad (1)$$

where

$$\alpha = N_2/N_1,$$

N_1 and N_2 = the stress resultants along the major and minor composite laminate axes,

D_{11} , D_{22} , D_{12} , and D_{66} = bending stiffnesses of the composite laminate,

a = the dimension of the square plate (19 mm), and

n and m = integers determining the buckling modes of the plate

Both m and n were found to be equal to one for the lowest buckling stress. Another two models were infinitely long strips along either one of the two loading axes. The widths of these strips

were equal to the distance between edges of two adjacent Teflon tabs (10 mm). Compressive loadings were applied along one axis, and two opposite sides of the strips were assumed to be simply supported. The buckling stress for the second model was calculated as

$$N_1 = - \frac{\pi^2 D_{11}}{b^2} \tag{2}$$

and the third model as

$$N_2 = - \frac{\pi^2 D_{22}}{b^2} \tag{3}$$

where *b* is the width of the strips. This calculation provided a simple estimate for the buckling stresses in which the directional dependence of the stiffness was considered. These three models were considered to be conservative in comparison with buckling condition of the specimen because edges of the plate were assumed to be simply supported while the laterally supported areas of the specimen were closer to a clamped condition. The bending stiffnesses were computed from the measured elastic constants to be *D*₁₁ = 48.6 N-m, *D*₂₂ = 12.4 N-m, *D*₁₂ = 6.8 N-m, and *D*₆₆ = 12.4 N-m for unidirectional composites and *D*₁₁ = 38.4 N-m, *D*₂₂ = 22.6 N-m, *D*₁₂ = 6.8 N-m, and *D*₆₆ = 12.4 N-m for cross-ply laminates. Results of this analysis are compared to the experimental data in the next section.

According to the buckling analysis, a total of 36 lateral supports were required on each side of the specimen for unidirectional or cross-ply E-glass/epoxy composite laminates. Every support was carefully adjusted to ensure its contact with the specimen surface. In addition, strain gages were mounted on the specimen surface before the installation of these lateral supports. The amount of strain in the specimen was monitored during the adjustment of these supports to minimize initial deflection of the specimen.

Results and Discussion

Using the in-plane biaxial testing system, a series of tests were conducted with E-glass/epoxy composites. Tables 1 and 2 are experimental results for the unidirectional and cross-ply laminates. In these tables, the failure stresses are the average stresses calculated from the maximum loads reached during the test. The biaxiality ratios are the ratios of displacements along major and minor material axes. These experimental data represent the failure criteria of E-glass/epoxy composite laminates under in-plane biaxial compressive stress. Figures 6 and 7 show the failure criteria in a two-dimensional stress plane. The failure envelopes are nearly rectan-

TABLE 1—Experimental data resulting from the in-plane biaxial compression tests for the unidirectional E-glass/epoxy composite laminates

Specimen ID	Displacement Ratio, $\frac{d_2}{d_1}$	$-(\sigma_u)_1$, MPa	$-(\sigma_u)_2$, MPa
U01	0	439	0
U02	1.0	208	115
U03	0.5	301	94
U04	∞	0	99
U05	0.2	412	70
U06	2.0	61	100
U07	0.25	396	91

TABLE 2—Experimental data resulting from the in-plane biaxial compression tests for cross-ply E-glass/epoxy composite laminates.

Specimen ID	Displacement Ratio, $\frac{d_2}{d_1}$	$-(\sigma_u)_1$, MPa	$-(\sigma_u)_2$, MPa
CO1	0.5	293	185
CO2	2.0	153	248
CO3	0.2	321	54
CO4	3.0	71	269
CO5	0.33	350	123
CO7	0.5	343	220
CO8	0	356	0
CO9	0.7	335	217
C10	∞	0	266
C11	1.0	263	260

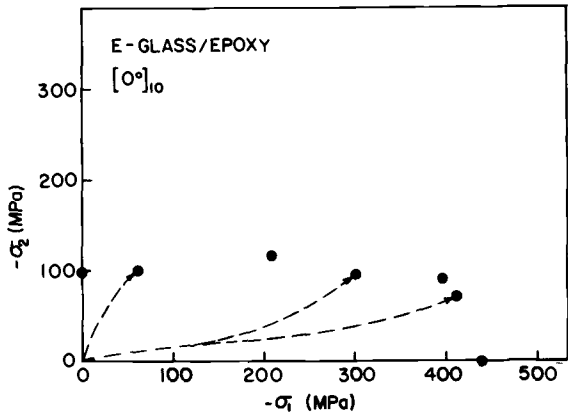


FIG. 6—In-plane biaxial compressive strength of E-glass/epoxy laminate—unidirectional.

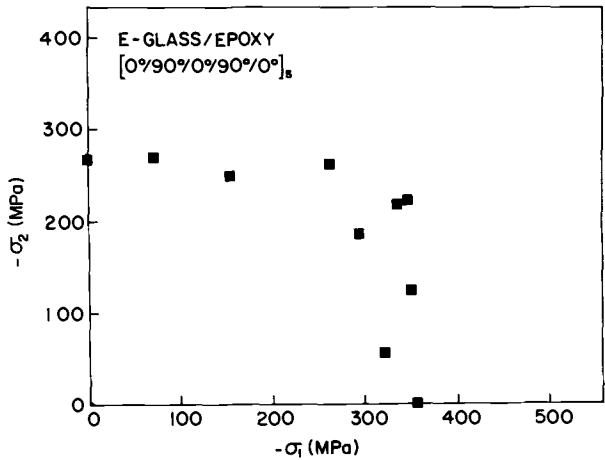


FIG. 7—In-plane biaxial compressive strength of E-glass/epoxy laminate—cross ply.

gular. However, the data seem to indicate a small influence of biaxiality on failure strengths when both stresses are near the uniaxial strength.

Because all tests were conducted with displacement control, the stress ratios were not constant throughout the loading history. Figure 6 also shows the typical stress histories for the unidirectional composite laminate. Proportionality of the loading histories changes significantly either at the beginning or close to the end of tests. The stress nonproportionality at the beginning of tests may result from the seating in the loading fixture. The nonproportionality close to the end of tests results from the material nonlinearity before failure of the specimen. It is still unclear how this stress nonproportionality may affect the failure strength of composite materials.

Movement of the linear actuators was stopped when the specimen load decreased by 5% indicating specimen failure. The broken surface of the failed specimen was maintained for the study of failure mechanisms. Figures 8 and 9 are photographs of the broken specimens. Failure

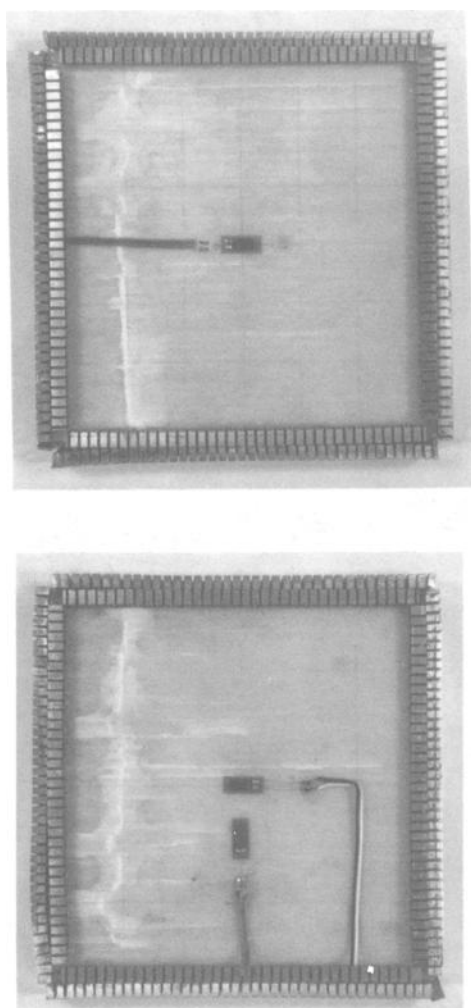


FIG 8—Unidirectional test specimens after failure (top) $\sigma_1 = -411 \text{ MPa}$, $\sigma_2 = -70 \text{ MPa}$ and (bot-

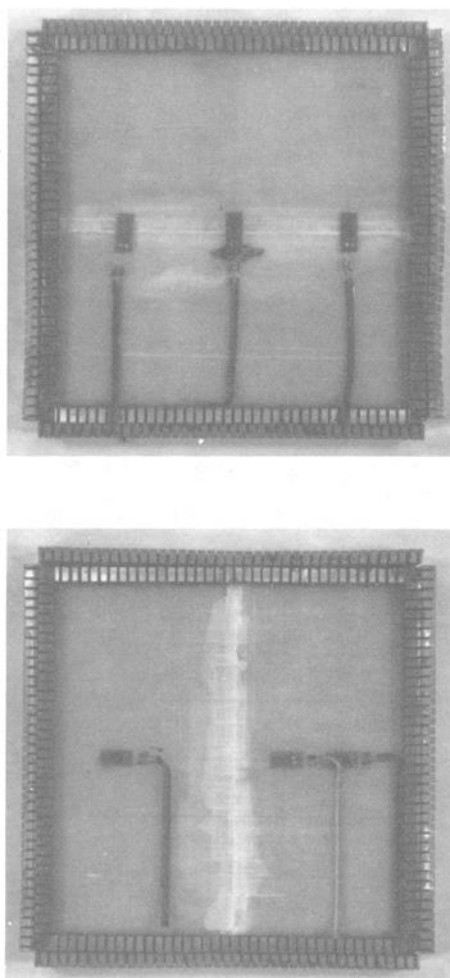


FIG 9—Cross-ply test specimens after failure (top) $\sigma_1 = 0 \text{ MPa}$, $\sigma_2 = -263 \text{ MPa}$ and (bottom) $\sigma_1 = -350 \text{ MPa}$, $\sigma_2 = -123 \text{ MPa}$

occurs inside the test section for all the specimens tested. Depending on the biaxiality stress ratios, the broken surfaces are perpendicular to either one of the two loading axes. For unidirectional laminates, specimens appear to fail by shear stress through the thickness direction. For cross-ply laminates, a significant amount of interlaminar delamination occurs although, macroscopically, failure also appears to result from the out-of-plane shear stress. A detailed microscopic analysis is in progress, and results will be reported in the future.

Although data have been obtained with this test method, the experimental results are reliable only if the stress in the specimen is homogeneous. Stress inhomogeneity may result from two sources. The first is due to the friction stress between the specimen edges and the steel blocks (see Fig. 3) and the friction between the specimen surface and the lateral supports (see Fig. 4). Stress inhomogeneity resulting from this source occurs along in-plane direction of the

specimen. Another possibility is that the specimen can buckle before the compressive stress reaches the material strength. In-plane stresses vary along the thickness direction of the specimen when bending occurs. The following discussion focuses on the stress homogeneity and the buckling condition of the specimen. The material strengths obtained from this testing method are also compared to the results from the standardized unidirectional test method. Accuracy of the experimental data from this biaxial testing method is discussed.

Stress Homogeneity

To examine the stress homogeneity, three strain gages were used for measuring strain in the specimen. These strain gages were mounted on the same side of the specimen. One gage was at the center of the specimen; the other two gages were about 25 mm away from the specimen edges. Figure 10 shows the typical loading and unloading stress-strain curves recorded from these gages together with the positions of the strain gages. Results are only shown for the cross-ply laminate. Similar results were obtained for both unidirectional and cross-ply laminates. In this case, compressive loading was applied along one axis. Because three stress-strain curves overlap each other throughout the entire loading history, stress should be homogeneous under this loading condition.

The same measurements were made for specimens under biaxial loading. Figure 11 shows the typical stress-strain curves for these experiments. The strain is not homogeneous and has the highest value at the center of the specimen. As in the case of unidirectional loading, similar results were obtained for both unidirectional and cross-ply laminates during axial loading. The strain close to the specimen edges is about 20% less than that at the center. Because the specimen edges are connected to rigid blocks for load transfer (see Fig. 3), these results suggest that the current specimen design is not able to eliminate the end effects. Also, it has been found that the strain distribution is not symmetrical to the center line of the specimen. According to the experimental data, the difference of strain between G2 and G3 is about 0.002. For a 127-mm-long specimen, this corresponds to about a 0.2-mm difference in the displacement between two sides of the specimen. This difference is close to the tolerance specified for machining the specimen fixture. A systematic study of the alignment is needed to establish the causes of the nonuniform strain distribution.

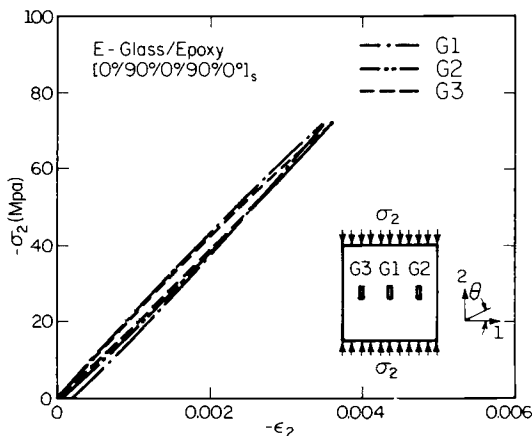


FIG. 10—Stress-strain response of biaxial specimen—unidirectional.

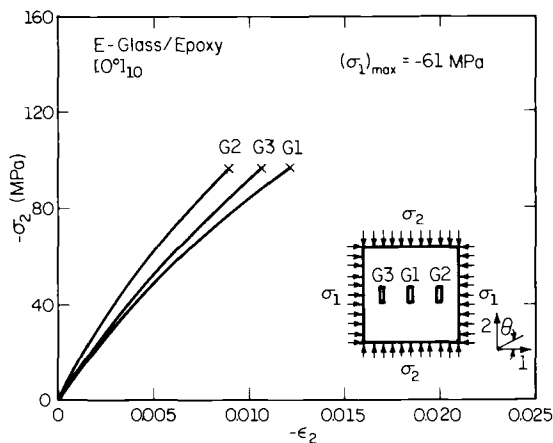


FIG. 11—Stress-strain response of biaxial specimen—cross ply.

Buckling Condition

The critical buckling stresses of the composite specimen have been estimated from Eqs 1 through 3. These stresses are compared to the experimentally obtained material strengths in Fig. 12 for unidirectional composite laminates and in Fig. 13 for cross-ply laminates. For both cases, the compressive failure strengths are less than the buckling stresses of the specimen. This means that, ideally, the composite specimen should not buckle before it reaches the compressive failure strength. However, this conclusion is based on the assumptions that the composite material does not exhibit any material nonlinearity and that the lateral supports are able to prevent deflection. These assumptions cannot be guaranteed in the tests, and the buckling condition of the specimen has to be examined experimentally.

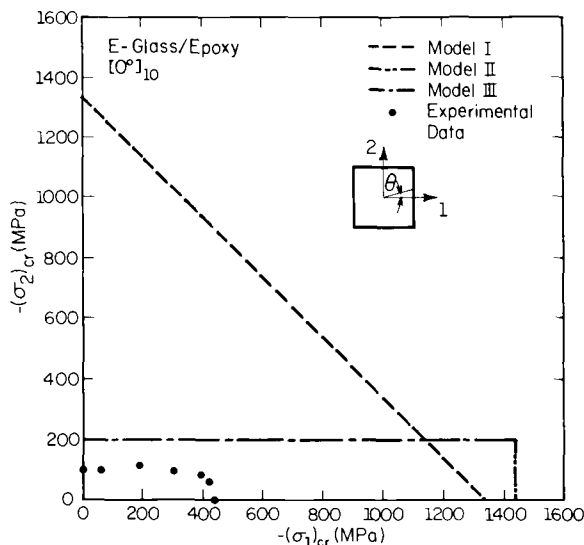


FIG. 12—Buckling failure envelope for test specimen—unidirectional.

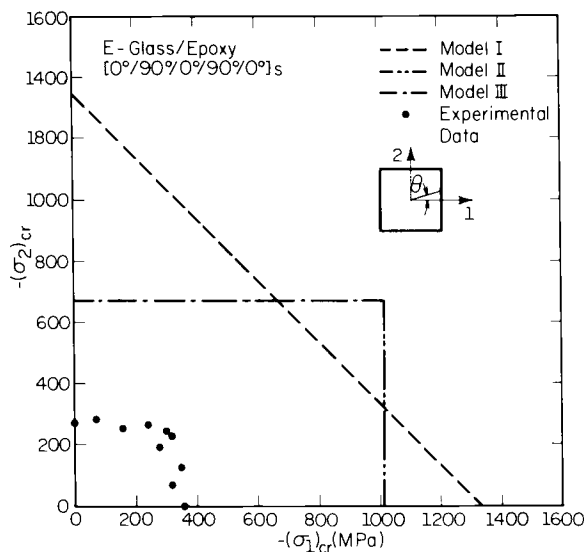


FIG. 13.—Buckling failure envelope for test specimen—cross ply.

To study the buckling condition of the specimen, two strain gages were used to measure the amount of strain for each loading axis. These two strain gages were mounted on different sides of the specimen. If the specimen starts to buckle, the strain measured from these two gages will be different. Figure 14 shows the stress-strain curves for unidirectional laminates under biaxial loading in which both σ_1 and σ_2 are simultaneously increased to failure. In this figure, G1 and G2 are strain gages along the major composite laminate axis and G3 and G4 are gages along the minor composite laminate axis. The stresses and strains in Figs. 14 and 15 for G1 and G2 represent σ_1 and ϵ_1 . The stresses and strains for G3 and G4 represent σ_2 and ϵ_2 . Results from this experiment show that the specimen buckles slightly along the fiber direction of the spec-

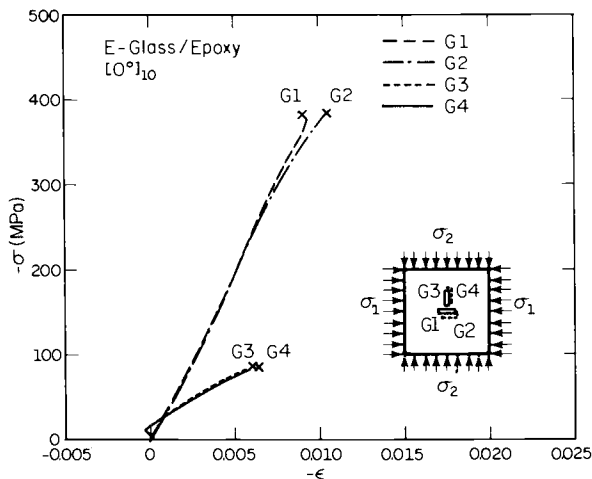


FIG. 14.—Stress-strain response for buckling test—unidirectional.

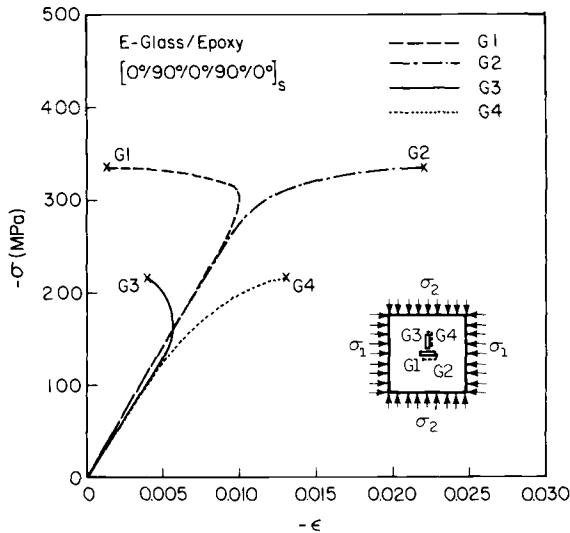


FIG. 15—Stress-strain response for buckling test—cross ply.

imen before reaching the failure stress. The same experiment has been conducted for the cross-ply laminate (see Fig. 15). In this case, the specimen buckles significantly along two axes before it fails. It is still unclear whether buckling is due to damage in the specimen or structural buckling because of the spacing of the lateral supports. Examination of the failed specimens shows a shear failure through the thickness of the specimen. The apparent buckling measured by the strain gages could be caused by a shear failure in another portion of the specimen.

Uniaxial Test Method Comparison

The experimental data resulting from the biaxial test method have been compared to those from the Illinois Institute of Technology Research Institute (IITRI) test method (ASTM Test for Compressive Properties of Unidirectional or Crossply Fiber-Resin Composites [D 3410]). To examine the buckling condition of the specimen, strain gages were mounted on both sides of the uniaxial specimen. Figures 16 and 17 show the stress-strain curves obtained from both methods. Because the compressive failure strengths of E-glass/epoxy composite laminates are found to be relatively insensitive to the stress biaxiality, biaxial data for the buckling tests are compared to the uniaxial data for the IITRI test.

Two major differences between the data from the biaxial test method and the IITRI test method were observed. First, the compressive strength obtained from the biaxial test method is much lower than that from the IITRI method. For a unidirectional composite laminate, the difference is about 25%, whereas, for the cross-ply laminate, it can be as high as 70%. The failure strength of composite materials is known to decrease as the volume of material increases [6]. The second difference between the two specimens is in the buckling conditions. Based on the stress-strain curves in Fig. 16, the IITRI specimen appears to have an initial curvature before testing. The specimen is laterally deflected throughout the test. However, the small gage length and tightly clamped edges of the IITRI specimen allow it to endure a higher stress than the relatively large biaxial specimen. The same tendency can also be found for the cross-ply laminates (see Fig. 17). These differences indicate the importance of the boundary conditions of the specimen for compressive testing of composite laminates. Both specimens

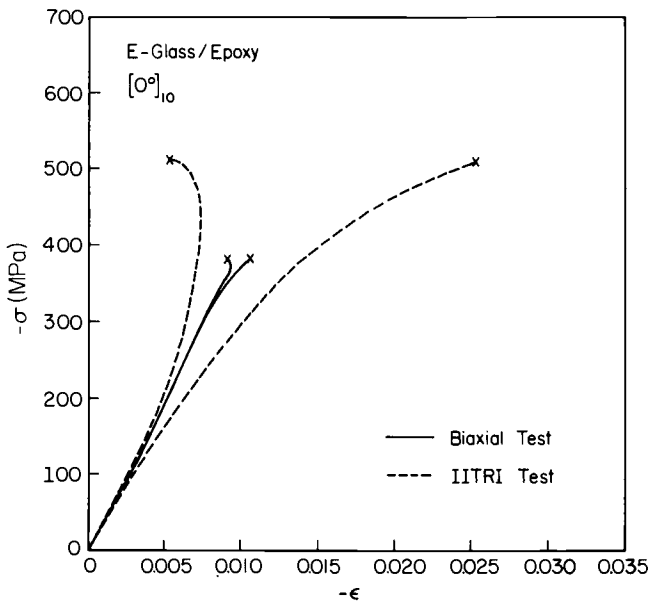


FIG. 16—Comparison of stress-strain response from biaxial and IITRI test specimens—unidirectional.

would be expected to have higher strengths if the thickness were increased to prevent buckling failures.

A detailed study of the experimental data from all of the biaxial tests shows that the stress distribution inside the specimen is relatively homogeneous. A 20 to 30% strain variation was

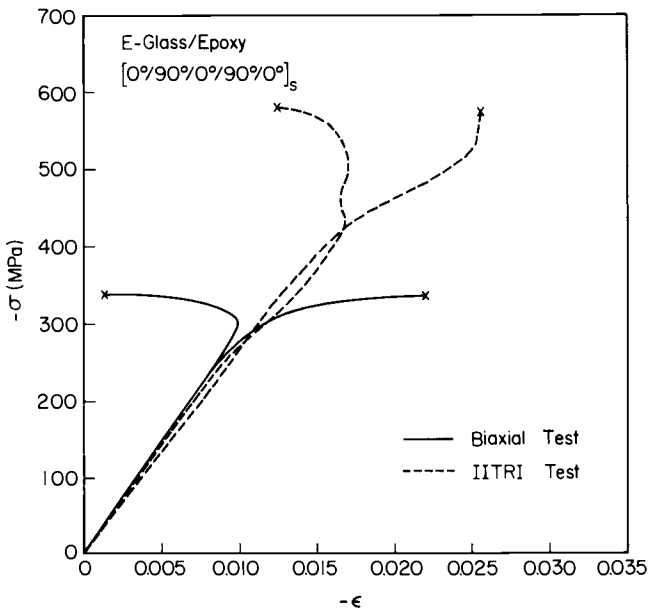


FIG. 17—Comparison of stress-strain response from biaxial and IITRI test specimens—cross ply.

observed in the biaxial specimen with the highest strain (stress) occurring at center of the specimen. This stress variation is comparable to the level of stress concentration in many uniaxial specimens [7]. Failure of the biaxial specimen occurs inside the test section and does not occur from a phase concentration at the corners. This is a significant feature of this method because failure has often been found inside or close to the grip areas for other test methods. The buckling condition of the biaxial specimen is a critical issue requiring more attention. The number and the arrangement of the lateral supports must be changed to justify the relatively low failure strengths obtained from this method. This modification is important in determining whether the low failure strength of the biaxial specimen is due to the large specimen size or the boundary conditions of the test specimen. Microstructural examination of the failed specimens is also necessary to clarify the failure mechanisms. The lower failure strengths could also be due to an increase in degree of delamination observed in the larger specimen after testing. This would effectively reduce the specimen thickness and cause premature buckling.

Summary and Conclusion

An in-plane biaxial compressive test method has been developed for composite laminates. This experimental technique has been used to determine the compressive failure strengths of E-glass/epoxy composite laminates with different lamination variables and stress biaxiality ratios. The failure strength determined from this new method is lower than that obtained from the uniaxial IITRI method, and the compressive strengths are found to be relatively insensitive to the stress biaxiality. A detailed study of the testing system shows that the stress distribution is relatively homogeneous for the biaxial specimen. The buckling condition of the biaxial specimen is significantly different from that of the IITRI specimen.

Acknowledgment

The work reported in this paper was supported by the Office of Naval Research (ONR), Arlington, VA, through Grant N00014-86-K-0467 to the University of Illinois.

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The Effect of Tab Orientation on the Distribution of Strains in Composite Specimens

REFERENCE: Foos, B C , Wolfe, W E , and Sandhu, R S , "The Effect of Tab Orientation on the Distribution of Strains in Composite Specimens," *Composite Materials Testing and Design (Tenth Volume)*, ASTM STP 1120, Glenn C Grimes, Ed , American Society for Testing and Materials, Philadelphia, 1992, pp 103–113

ABSTRACT: An accurate measurement of tensile strength of graphite/epoxy composites can be difficult because of nonuniformities in the stress/strain field. To improve the uniformity of the strain state in 0° , $\pm 45^\circ$, and $(0^\circ/\pm 45^\circ/90^\circ)_s$ laminate specimens, the design of end tabs was studied using the finite element technique. The results of the analyses indicate that the use of tapered chevron tabs with chevron angles of 2, 25, and 5°, respectively, for the three laminates, yield more uniform strain states than the square-ended tapered tabs. In the case of $\pm 45^\circ$ laminates with the tabbing material oriented at 45° , additional improvement in the strain state occurs with ends of the tapered tabs chevroned at 20° . Tests conducted on $\pm 45^\circ$ laminates with three types of tapered tab ends (namely, square ends and chevroned at 25 and 20°) verified the analytical findings.

KEYWORDS: composite materials, tension specimens, finite element analysis, end tab design, fiberglass

The material properties used in the design of structures are obtained in the laboratory using procedures in which specimens with simple boundary conditions are subjected to known load or displacement conditions. One of the most widely used tests for measuring the properties of composite materials is the coupon tension test. In this test, it is assumed that the simple specimen geometry and boundary conditions ensure that the longitudinal strains are uniform throughout the test section. However, when the material properties of composite laminates are evaluated, the typical coupon consists of not only the composite, but also end tabs bonded to the specimen that are, in turn, gripped by the loading machine to apply the desired test load. The tabs are used to make the load transfer from the grips to the specimen more uniform. The use of different materials for the tab, the adhesive, and the specimen causes a mismatch in stiffness and Poisson's ratio at the tab line. The properties' mismatch, in turn, results in nonuniformities in stresses and strains near the tab regions of the specimen. To minimize the effects of these nonuniformities on specimen elastic properties, minimum coupon lengths are generally specified and the measurement of material strain is made well away from the tab. Adjustments to the methods for making strength measurements cannot be accomplished as

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easily as those for elastic properties. If the coupon fails near the tab where the actual stress distribution is not known, precise strength determinations cannot be made. Therefore, a tab design that minimizes the stress and strain nonuniformities would improve the estimate of composite strength by increasing the reliability of critical laboratory test measurements.

The effects of end restraints on the uniformity of stresses in composite specimens loaded in tension have been studied by a number of investigators [1–12]. These references pertain to two types of specimens, namely, off-axis specimens and tensile coupons.

Pagano and Halpin [1] conducted analytical and experimental investigations of the effects of restraints caused by clamping devices on the response of off-axis specimens. Wu and Thomas [2] found that by permitting the ends of the specimens to rotate, the effects of clamping devices could be reduced. Rizzo [3] used the finite element technique to demonstrate the influence of rigid clamping with and without end rotations on the distribution of stresses in the test section of the composite specimen. Sandhu and Sendeckyj et al. [4,5], using linear finite element analysis, determined that a state of uniform axial stress within the test section could be achieved if (a) end grips were allowed to rotate and (b) the fiber angle of fiber glass tab material and the inclination of tapered tab ends were simultaneously optimized. Using nonlinear finite element analyses, they found that the specimens so designed maintained the uniformity of the stress state even in the nonlinear range. Cron et al. [6] showed that by adjusting the amount of the clamped area of tabs and appropriately locating the point of clamp rotation, a state of nearly uniform stress could be generated. Observing that undesirable stress concentrations still exist in the tab area even for tapered tabs, Sun and Berreth [7] suggested using a new type of tab material, one in which a fiber glass mesh was embedded in a silicon rubber matrix. The strains measured using this new tabbing material were reported by Sun and Berreth to be more uniform than those obtained in specimens bonded to the conventional fiber-glass tabs.

In an attempt to reduce stress concentrations in tensile coupons, Lenoe [8] embedded ends of the specimens in a tough epoxy that was machined to novel shapes. Reiss and Yao [9] studied the edge-constrained effects and the in-plane bending phenomenon caused by nonuniform gripping. Kural and Flagg [10] used the finite element method to analyze the distribution of stresses in unidirectional graphite/epoxy tensile coupons. Their work showed that tapering the tabs reduced nonuniformities in the stress field near the tab line. Cunningham et al. [11] conducted a finite element study and performed laboratory experiments on unidirectional graphite/epoxy specimens of varying tab geometries. They found that peak shear and normal stresses were related to tab length, taper, and thickness. They reported that increasing the length of the tab's taper could reduce nonuniformities in the specimen stress field. The tapered tab specimens, however, have a tendency to delaminate near the tabs. To reduce stress concentrations, Oplinger et al. [12] pursued an alternative approach by streamlining the width of the specimen to avoid the problems associated with testing of straight edge specimens.

In the study reported herein, we apply some of the concepts of Sandhu and Sendeckyj [4,5] to the tension specimens with tabs. Initially for this purpose two candidate specimens, one with tapered tabs and the other with tabs without any taper, were selected. However, preliminary investigations determined that specimens with tapered tabs had lower stress concentration than the specimen without tapered tabs. Consequently, only the specimens with tapered tabs and the following laminates were included:

- (a) 0° laminates,
- (b) $\pm 45^\circ$ laminates, and
- (c) $0^\circ/\pm 45^\circ/90^\circ$ laminates.

Analysis

To evaluate concentration of axial stresses and strains, a two-dimensional (2-D) finite element program (PLSTR) that uses quadrilateral elements was developed at the Wright Laboratory. This program uses a linear elastic ply-by-ply analysis that takes into account the number and orientation of the individual plies in the formulation of laminate properties. For assessing the analytical results in the study, we define the axial strain difference as the maximum axial strain minus the minimum axial strain in the composite coupon as the measure of the uniformity of strain field (that is, the smaller this value, the more uniform the strain state). Following the concepts of Refs 4 and 5, the principal parameter is the inclination angle λ of ends of tapered chevron tabs.

Finite Element Analysis for $\lambda = 0^\circ$

The specimen with square-ended tapered tabs along with the finite element (FE) model that corresponds to the top left quarter of the specimen is shown in Fig. 1. The model consists of

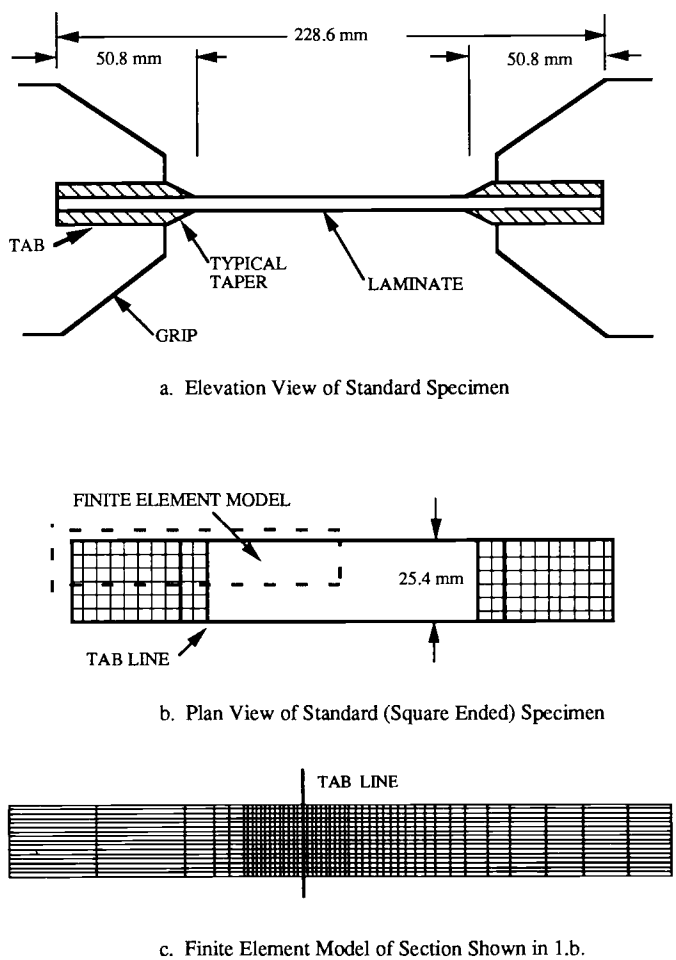


TABLE 1—Elastic moduli of graphite/epoxy, tab material, adhesive, and grip

Material	E_{11} , GPa	E_{22} , GPa	G_{12} , GPa	μ_{12}
Graphite/epoxy	138 590	10 343	4 551	0 290
Tab Material	32 544	32 544	6 895	0 077
Adhesive	2 186	2 186	0 758	0 420
Grip	200 000	200 000	77 224	0 30

816 nodes and 3782 stacked elements. In the tab area, the FE model includes elements for the gripping steel wedges, the fiberglass tabs, the adhesive bond line (0.127 mm thick), and the composite laminate. The material properties (extracted from Ref. 10) used in the finite element analysis are shown in Table 1. At the right end of the model, all nodes are subjected to a uniform longitudinal displacement of 0.1524 mm. The data obtained from the analyses are shown in Table 2. As expected, the influence of tabs on the distribution of strains is minimal for 0°/90° and 90° laminates. The laminates most affected by the “standard” tabs ($\lambda = 0^\circ$, taper = 20°, glass fibers oriented at 0°/90°) are the $\pm 45^\circ$ laminates.

Finite Element Analysis for $\lambda \neq 0^\circ$

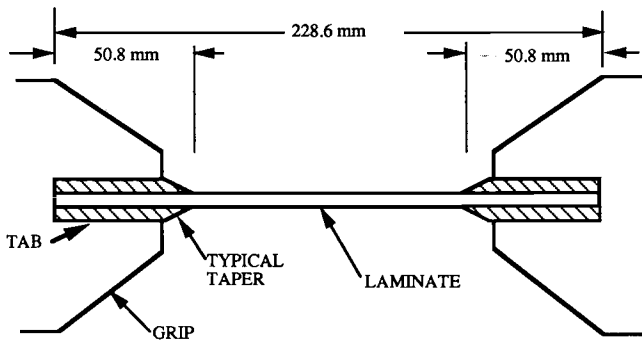
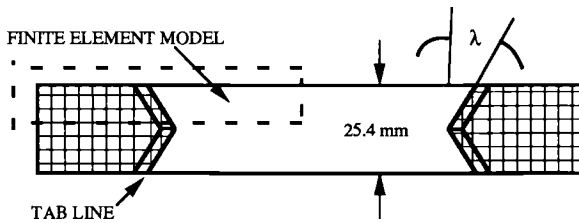
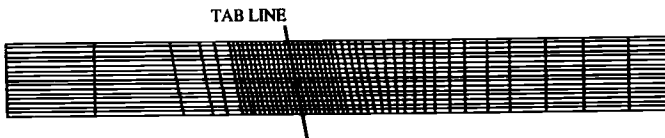
Using the technique of Refs. 4 and 5, square-ended tapered tabs were cut at a chevron angle λ . The specimen corresponding to the cut along with the finite element idealization is shown in Fig. 2. The model shown in Fig. 2 also consists of 816 nodes and 3792 stacked elements similar to the model of Fig. 1. The nodes at the right side were displaced by 0.1524 mm in the longitudinal direction. For various values of the parameter λ , a series of finite element analyses for 0°, $\pm 45^\circ$, and (0°/ $\pm 45^\circ$ /90°)_s laminates with standard fiberglass tabs was conducted. In addition, we studied the effect of changing the orientation of the fibers in the tabbing material by $\pm 45^\circ$ for various values of λ . The data obtained from the finite element analyses are shown in Fig. 3. The plots indicate that the axial strain difference tends to decrease with increasing tab chevron angle λ , up to a point, and thereafter it tends to increase. The inclinations that correspond to the minimal values of the axial strain differences, λ_{min} , for 0°, $\pm 45^\circ$, and (0°/ $\pm 45^\circ$ /90°)_s laminates with standard tabs in Fig. 3 are observed to be 2°, 25°, and 5°, respectively. The minimum value of λ for the $\pm 45^\circ$ laminates with the tab material oriented at $\pm 45^\circ$ is 20°.

Effect of E_{11} , E_{22} , and G_{12} of Lamina on λ_{min} of $\pm 45^\circ$ Laminate with Standard Tabs

To determine the sensitivity of λ_{min} to the changes in the elastic properties of the laminates, we selected $\pm 45^\circ$ laminates with the tabs attached in a “standard” configuration. In a parametric study, combinations of the elastic properties (given in Table 3) were devised to perform a series of finite element analyses with varying λ . The results of this investigation are plotted

TABLE 2—Axial micro strain differences in laminates for $\lambda = 0$ and 0.1524-mm displacement

Laminate	0°	$\pm 45^\circ$	(0°/ $\pm 45^\circ$ /90°) _s	90°	0°/90°
Axial Strain Difference	62.13	1886.82	871.15	29.32	12.30

a. Elevation View of Specimen, Chevron Angle λ b. Plan View of Specimen, Chevron Angle λ 

c. Finite Element Model of Section Shown in 2.b.

FIG. 2—Specimen with chevroned tabs.

in Fig. 4. From the figure, we find that within the range 110.32 to 165.48 GPa for E_{11} , $\lambda_{\min}$ remains constant at 25° . Similarly, there is no change in $\lambda_{\min}$ when E_{22} assumes values from 6.895 to 13.79 GPa. However, for cases in which E_{11} and E_{22} are held constant and G_{12} is varied, $\lambda_{\min}$ tends to decrease with increasing values of G_{12} .

Strain Contours and Deformed Shapes

Strain Contours—Using the finite element data, axial strain contours of $(0^\circ/\pm 45^\circ/90^\circ)_s$ and $\pm 45^\circ$ laminates were generated. These axial strain contours near the tab areas are shown in Figs. 5 and 6, respectively. Figure 5a is a contour plot for $(0^\circ/\pm 45^\circ/90^\circ)_s$ laminates with square-ended tabs ($\lambda = 0^\circ$). It is marked by the existence of very high axial strain gradients

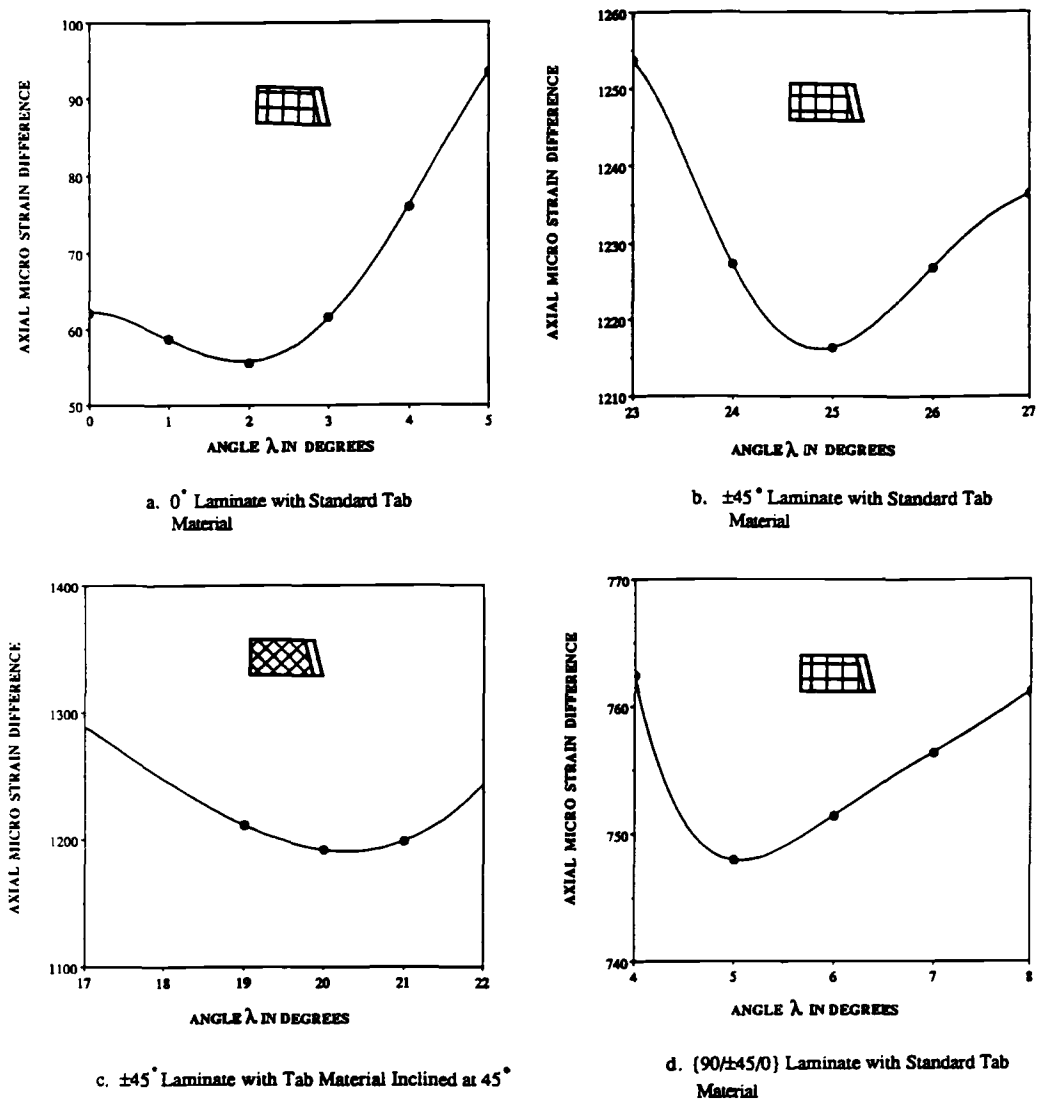


FIG. 3—Effect of the inclination of tabs on the axial strain distribution in different laminates.

along the tab line ($\epsilon_{x_{\max}} = 2400 \mu\epsilon$). In the case of the laminates with inclined tabs (Fig. 5b), axial strain gradients near the top left corner are less severe ($\epsilon_{x_{\max}} = 2260 \mu\epsilon$). A similar variation of axial strain gradients for $\pm 45^\circ$ laminates can be observed in Fig. 6a and Fig. 6b. In the case of $\pm 45^\circ$ laminates with the tab material oriented at $\pm 45^\circ$, additional improvement occurs ($\epsilon_{x_{\max}} = 2600 \mu\epsilon$ for $\lambda = 0^\circ$, $\epsilon_{x_{\max}} = 2300 \mu\epsilon$ for $\lambda = 25^\circ$, $\epsilon_{x_{\max}} = 2000 \mu\epsilon$ for $\lambda = 20^\circ$ and tab fibers $= 45^\circ$).

TABLE 3—Elastic lamina moduli E_{11} , E_{22} , G_{12} , and μ_{12} of graphite/epoxy.

Material	E_{11} , GPa	E_{22} , GPa	G_{12} , GPa	μ_{12}
I	138.590	10.343	4.551	0.29
II	110.320	10.343	4.551	0.29
III	165.480	10.343	4.551	0.29
IV	138.590	6.895	4.551	0.29
V	138.590	13.790	4.551	0.29
VI	138.590	10.343	5.516	0.29
VII	138.590	10.343	6.895	0.29

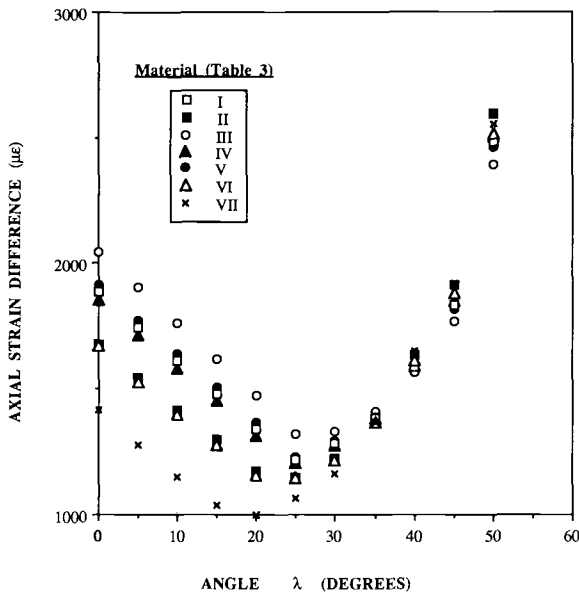


FIG. 4—Effect of varying E_{11} , E_{22} , and G_{12} on axial strain distribution of $\pm 45^\circ$ laminates.

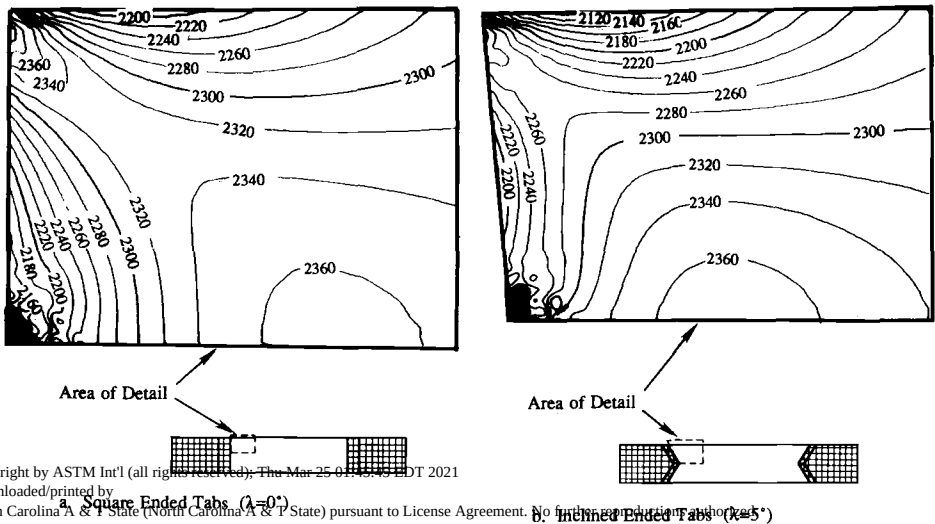


FIG. 5—Strain contours of $(0^\circ/\pm 45^\circ/90^\circ)_s$ laminates.

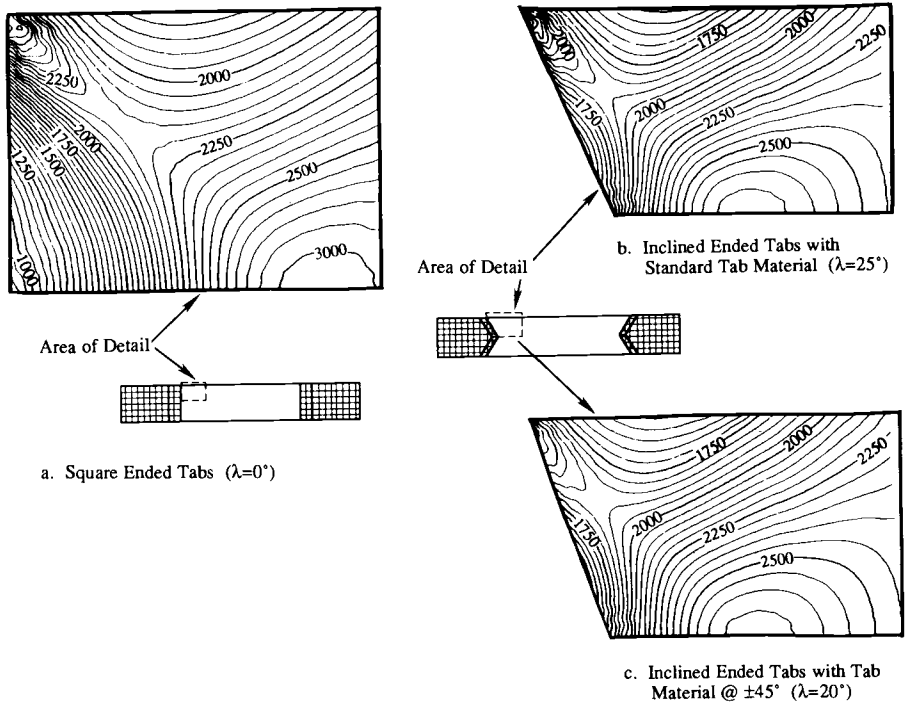


FIG. 6—Strain contours of $\pm 45^\circ$ laminates.

Deformed Shapes—In Fig. 7, a plot of the deformed shape of the $(0^\circ/\pm 45^\circ/90^\circ)_s$ composite coupon beginning at the tab line and moving for approximately 75 mm along the free edge is presented for the two tab geometries studied. A similar plot is shown in Fig. 8 for the three tab geometries analyzed for the $\pm 45^\circ$ laminates. What can be seen in each of these figures is a more uniform transition in the transverse strain from the tab area where lateral deformation is prevented because of the presence of the tab and the grip, to the “free field” where tab effects become minimal. The analytical results depicted in these figures offer additional support to the observation that strain gradients are reduced when tab orientation is modified.

Experimental Observations

Specimens were fabricated using $\pm 45^\circ$ layup with square-ended tabs, chevroned tabs with standard tabbing material (that is, glass fiber directions = $0^\circ/90^\circ$), and chevroned tabs with tabbing material oriented at $\pm 45^\circ$. The sample size in each of the categories consisted of five specimens. These specimens were tested at 1.27 mm/min in an Instron Testing Machine. Photographs of the specimens tested to failure are presented in Fig. 9. The specimens with square-ended tabs had an average strength at failure of 152.9 MPa. The specimens with chevroned tabs and standard tabbing material failed at an average stress of 161.9 MPa, for an increase of strength of 5.8%. Additionally, although the sample size was small, a distinct improvement in the location of the specimen failure was observed. One of the five specimens with square-ended

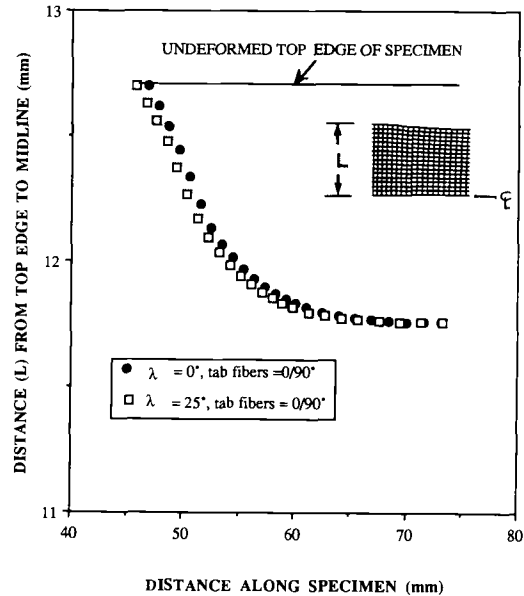


FIG. 7—Deformed and undeformed top edge of $(0^\circ/\pm 45^\circ/90^\circ)$ laminates.

tabs failed near the tabs. On the other hand, the specimens with chevroned tabs and standard tabbing material all tended to fail away from the tabs. Also, two of the specimens with the modified tabs appear to have failed at several locations within the central portion of the specimen, indicating a broad region of essentially uniform stresses and strains.

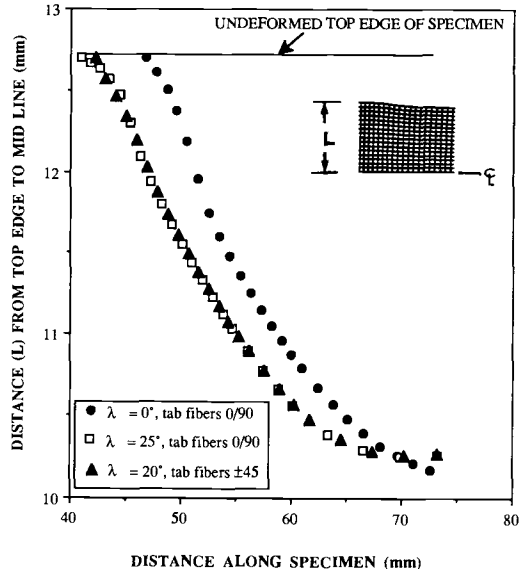


FIG. 8—Deformed and undeformed top edge of $\pm 45^\circ$ laminates.

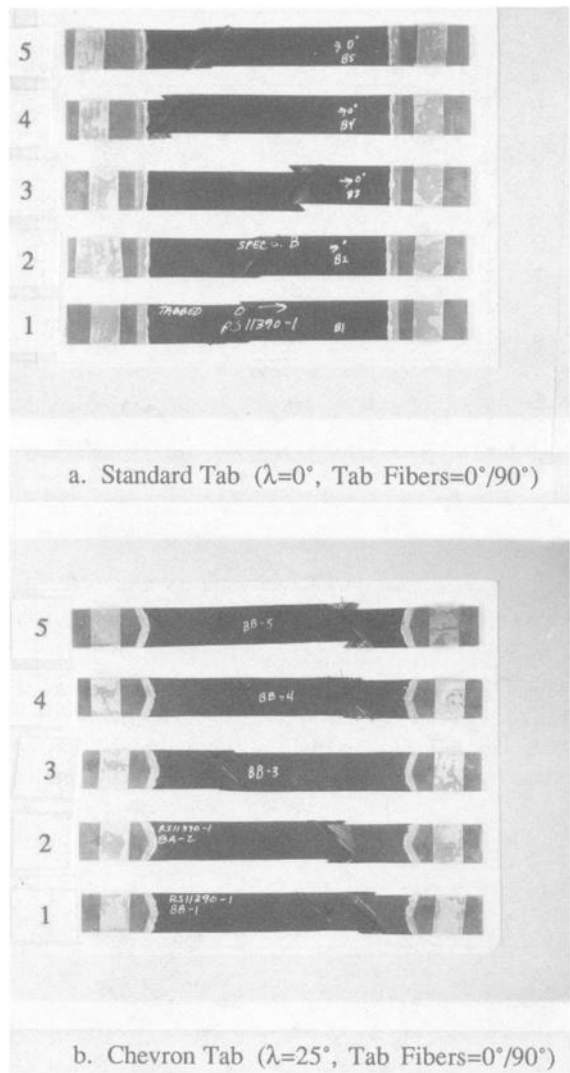


FIG. 9— $\pm 45^\circ$ tension test specimens.

Conclusions

The investigations have shown the beneficial effect of the tab end orientation of tapered tabs in improving the uniformity of the strain state in composite tensile coupons. By the use of standard fiber glass tabs with ends inclined at 2, 25, and 5° for 0, ± 45 , and $(0^\circ/\pm 45^\circ/90^\circ)$, laminates, respectively, the adverse strain gradient effect of square-ended tabs was shown in the finite element analysis to be reduced. In the case of $\pm 45^\circ$ laminates, the strain field can further be improved by simultaneously orienting the tabbing material at $\pm 45^\circ$ and choosing a tab chevron of 20°. A comparison of the results of laboratory tension tests on specimens having tapered tabs oriented with the tab line inclined ($\lambda = 25^\circ$) to the long axis of the speci-

men shows an increase in strength over the strengths obtained for specimens having tabs normal to the coupon's long axis ($\lambda = 0^\circ$), validating the analytical results oriented during this study. It is clear that the use of chevroned tabs with $\pm 45^\circ$ laminates will result in more failures occurring away from the tabs and thereby improve the shear data obtained using $\pm 45^\circ$ laminates.

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Session III

Material Mechanical Properties and Failure Criteria

Stephen R. Swanson, Chairman

Overview

Mechanical properties and failure criteria are the technical elements necessary for developing good design properties. Such properties are generated on IM7/PEEK, the experimental problems and solutions of hot-wet testing of graphite/polyimide laminates are delineated, and a new failure criteria for laminate strength prediction is detailed in the papers in this session.

Three papers from this session are included in this STP. The first is "Characterization of IM7 Graphite/Thermoplastic Polyetheretherketone (PEEK) for Spacecraft Structural Applications" by Silverman, Wiacek, and Griese. This paper presents an evaluation of IM7/PEEK for mechanical properties particularly associated with spacecraft applications. The mechanical properties evaluated included 0 and 90° tension and compression, shear, flexural, coefficient of thermal expansion, outgassing, moisture uptake, and moisture expansion coefficient. A comparison was made with IM7/8551-7 thermoset matrix material.

The second paper is "Hot/Wet Testing of Celion 3000/PMR-15 Coupon Specimens," by Blount. This paper draws attention to the difficulty of performing mechanical property evaluations under hot/wet conditions without excessive loss of moisture in the specimen. Attention is also given to suitable strain-gaging techniques for these test conditions.

The third paper is "A Scientific Approach to Composite Laminate Strength Prediction" by Hart-Smith. This paper discusses fundamental issues in the strength prediction of fiber composite laminates and presents a failure theory for strength evaluation. In the paper, the need for considering different failure criteria for the different possible failure modes in laminates, identified as fiber, matrix, and interface failure is pointed out. The paper concludes that fiber shear strain is the critical parameter for fiber-dominated laminate failure and presents comparisons with biaxial test data from the literature.

These technical papers presented illustrate the importance of solving the testing problems to get good allowables and the subsequent use of these allowables with appropriate analytical methods and failure criteria in the design process.

Edward M. Silverman,¹ Chris R. Wiacek,¹ and Richard A. Griesel¹

Characterization of IM7 Graphite/ Thermoplastic Polyetheretherketone (PEEK) for Spacecraft Structural Applications

REFERENCE: Silverman, E. M., Wiacek, C. R., and Griesel, R. A., "Characterization of IM7 Graphite/Thermoplastic Polyetheretherketone (PEEK) for Spacecraft Structural Applications," *Composite Materials Testing and Design (Tenth Volume)*, ASTM STP 1120, Glenn C. Grimes, Ed., American Society for Testing and Materials, Philadelphia, 1992, pp. 118–130.

ABSTRACT: Graphite composites fabricated from thermoplastic polyetheretherketone (PEEK) IM7 graphite prepreg tapes were evaluated for the development of lightweight, strength-critical, dimensional stable spacecraft components. The objective of this investigation was to summarize the mechanical and physical properties of IM7 graphite/PEEK to compare its properties to that of similar graphite/epoxy composites.

A property database for thermoplastic composite materials was generated. Composite mechanical property characterization included 0° and 90° tensile and compressive, in-plane shear, and flexural properties. Physical property characterization included coefficient of thermal expansion, outgassing/condensable volatiles, and equilibrium moisture absorption at 71°C (160°F) and 95% relative humidity. Scanning electron microscopy (SEM) was used to investigate fiber/matrix distribution, fiber orientation, and fiber/resin interface adhesion. Results were compared with data on IM7 graphite/8551-7 epoxy laminates to assess the relative merits of the IM7 graphite/PEEK thermoplastic material system.

In general, IM7 graphite/PEEK thermoplastic composites possess a combination of properties that appear to match the requirements for several spacecraft applications. In addition to having very respectable structural properties, IM7/PEEK laminates exhibited lower moisture uptake and lower outgassing than IM7/8551-7 epoxy laminates. The equilibrium moisture gain for the thermoplastic composites was 0.18% compared to 0.80% for the epoxy composite system. SEM analysis indicated that the thermoplastic PEEK composite laminates exhibited excellent fiber/matrix distribution and interfacial fiber/resin adhesion. The PEEK thermoplastic matrix composite, therefore, provides a superior replacement to conventional epoxy matrix composites for spacecraft applications.

KEYWORDS: polyetheretherketone (PEEK), graphite fiber composites, unidirectional tape, mechanical properties, coefficient of thermal expansion, water absorption, coefficient of moisture expansion

Advanced spacecraft structural components require a combination of high specific strength and stiffness, design life greater than ten years, dimensional stability, and resistance to thermal cycle at temperatures from -157 to $+121^{\circ}\text{C}$ (-250 to $+250^{\circ}\text{F}$) in an environment that includes exposure to radiation. The current method of fabricating primary and secondary structural spacecraft components uses graphite/epoxy composites to achieve significant spacecraft weight reduction. However, their continued use is not expected to meet future dimen-

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sional stability and manufacturing requirements because of epoxy resin deficiencies. These include excessive moisture absorption, thermally induced microcracking of the epoxy resin, low toughness, and limited adaptability to cost-effective fabrication methods such as superplastic diaphragm forming and hot-transfer press molding.

Recent investigations [1-10] point to emerging graphite fiber-reinforced thermoplastic matrix composite materials, such as graphite/polyetheretherketone (PEEK), as superior replacements for epoxy-based composites. Expected performance improvements over current graphite/epoxy composites include enhanced damage tolerance [2], dimensional stability (that is, low moisture absorption/desorption) [6,9], minimum thermally induced microcracking [8], decreased fabrication cost [10], ability for reprocessability/reworkability, and indefinite prepreg shelf life.

Despite the advantages graphite/PEEK composites offer, very little development work has been conducted to evaluate their potential for use in space hardware applications. This can be attributed to the limited design databases, the requirement for fabrication methods that differ from those used for thermosetting composites, that is, high processing temperatures (343 to 416°C [650 to 780°F]), and the lack of prepreg tack and drape that makes layup and forming of complex shapes difficult. This paper summarizes the mechanical and physical properties of IM7 graphite/thermoplastic PEEK. The properties are discussed in terms of stiffness and strength, coefficient of thermal expansion, water absorption behavior, and coefficient of moisture expansion. The IM7/PEEK properties are compared with the properties of IM7/8551-7 epoxy composites.

Experimental Description

Materials

Aromatic polymer composites (APC-2) PEEK prepreg tapes reinforced with unidirectional IM7 graphite fibers were obtained from ICI Composites, Wilton, England. The prepreg is normally supplied at a ply thickness of 5.2 mils, the weight fraction of graphite fibers is 68% (60% fiber volume). The graphite fibers have a nominal modulus of elasticity of 280 GPa (40×10^6 psi) and a tensile strength of 5205 MPa (745 ksi). PEEK is a semicrystalline polymer with a relatively high glass transition temperature and melting point—144 and 342°C (291 and 648°F), respectively.

Fabrication of Laminates

All laminates were fabricated from unidirectional prepreg tapes using the novel superplastic diaphragm-forming process. In this method, unconsolidated thermoplastic preforms are placed flat over a male tool constrained within a cage assembly. A special superplastically formable diaphragm, such as aluminum or a high-temperature polyimide film (for example, Kapton) is placed over the preform, then the cage assembly is placed in a press where heat and diaphragm-sealing pressure are applied. To consolidate the layup, pressurized gas is introduced through the pressure port and air is drawn out the vacuum port. This method eliminates the vacuum bagging needed in autoclave consolidation and provides a more uniform pressure distribution than is possible with press consolidation. Figure 1 is a schematic of this process.

For laminate fabrication, eight to sixteen plies of prepreg tapes were sandwiched between a steel flat plate and caul sheet constrained within the superplastic diaphragm-forming cage assembly. The layup was heated at a rate of 3 to 6°C/min (5 to 10°F/min) to a temperature of 316°C (600°F), at which point pressure was gradually applied at a rate of 0.069 to 0.138 MPa (10 to 20 psi/min) up to a final pressure of 0.69 MPa (100 psi). The layup was heated at a rate

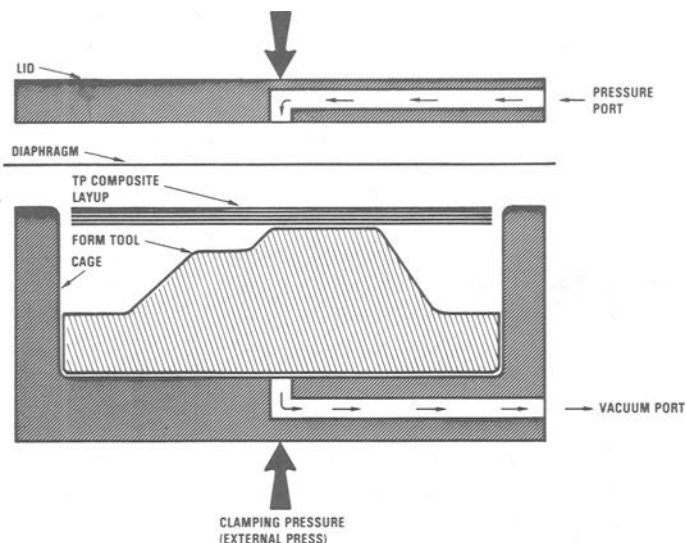


FIG. 1—Schematic of the superplastic diaphragm-forming process.

of 3 to 6°C/min (5 to 10°F/min) to the final melt temperature of $388 \pm 17^\circ\text{C}$ ($730 \pm 30^\circ\text{F}$). Forming temperatures were monitored by reading a thermocouple placed between the tool and part. The laminate was held at 0.69 MPa (100 psi) and the specified temperature for 10 ± 5 min before cooling. The superplastic diaphragm assembly was cooled by ambient water channeled through the platens. All laminates were cooled under pressure to below the glass transition temperature at a cooling rate of 3 to 4°C/min (6 to 8°F/min). The density of low-void content laminates was approximately 0.058 lb/in.³, or a specific gravity of 1.59 g/cm³.

PEEK in the presence of graphite fibers has a very forgiving morphology profile in which very small changes in percentage crystallinity have been observed over a very large range of cooling rates. Previous investigations [11–13] have measured the percentage crystallinity of the PEEK polymer in composites to be within a range of 30 to 35% for cooling rates varying within the range of 2 to 700°C/min (4 to 1260°F/min). If graphite/PEEK is cooled at rates less than 2°C/min (4°F/min), there is an increase in crystallinity, which reduces composite toughness. At cooling rates in excess of 700°C/min (1260°F/min), the growth of the polymer spherulites is incomplete and amorphous polymer remains between them. This may reduce stiffness and resistance to hostile solvents.

Laminate Quality

Scanning electron microscopy (SEM) was used to examine the quality of the composite laminates fabricated from prepreg tapes. Figure 2 shows the cross-sectional view of a polished graphite/PEEK specimen at different magnifications. The photomicrographs show good fiber/matrix distribution, no significant resin-rich areas, and no evidence of significant voids. Ultrasonic through-transmission records indicate that the laminates were practically void free.

Experimental Procedures

The fiber volume fractions of all laminates were measured by dissolving the thermoplastic matrix resin from the composites using the acid digestion procedure described by the ASTM

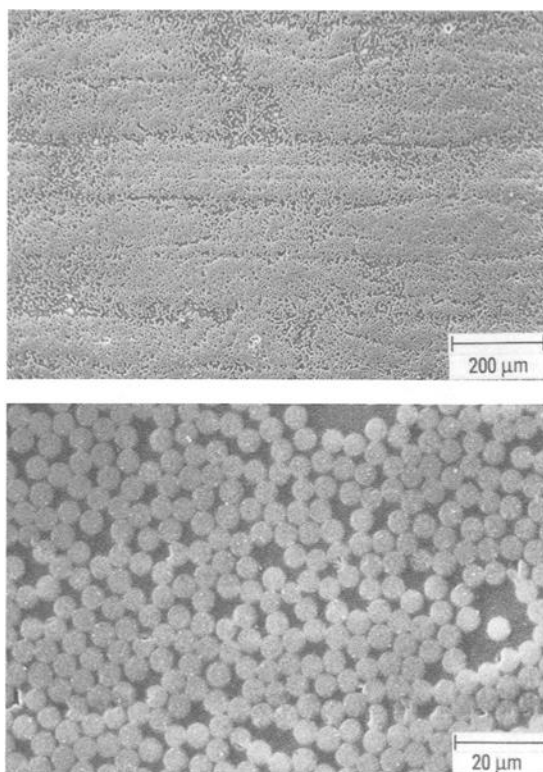


FIG. 2—Scanning electron micrographs showing the laminate quality of unidirectional specimens prepared from graphite/PEEK prepreg tapes.

Test for Fiber Content of Resin-Matrix Composites by Matrix Digestion (D 3171). The 0° tension testing was conducted according to the ASTM Test for Tensile Properties of Fiber-Resin Composites (D 3039), using a crosshead speed of 2.5 mm/min (0.1 in./min). Each 12-ply, 1.27- by 22.9-cm (0.5- by 9.0-in.) specimen contained a Poisson strain gage and glass fabric/epoxy tabs bonded to each end. Transverse tensile measurements were made to study the adhesion between the polymeric matrix and the graphite fibers. The 90° tension tests were conducted according to ASTM Test D 3039 on 16-ply, 16.5- by 2.54-cm (6.5- by 1-in.) specimens bonded with glass fabric/epoxy tabs and with one axial strain gage. Specimens were loaded at 1.27 mm/min (0.05 in./min). In-plane shear properties were determined from tension tests on $[\pm 45]_s$ laminates conducted according to the ASTM Practice for In-plane Shear Stress-Strain Response of Unidirectional Reinforced Plastics (D 3518). All specimens were of 22.9- by 2.54-cm (9- by 1-in.) configuration and bonded with glass fabric/epoxy tabs and one Poisson strain gage. Specimens were loaded at 2.50 mm/min (0.1 in./min). The 0 and 90° compression tests were conducted using the Celanese test method described by the ASTM Test for Compressive Properties of Unidirectional or Crossply Fiber-Resin Composites (D 3410), using a crosshead speed of 1.25 mm/min (0.05 in./min). The Celanese test fixture uses a 16-ply specimen of 141-mm (5.5-in.) length and 6.4-mm (0.25-in.) width configuration, bonded with glass fabric/epoxy tabs and two axial strain gages to measure modulus. Flexural properties were determined using the three-point loading method of the ASTM Test for Flexural Properties of Unreinforced and Reinforced Plastics and Electrical Insulating Materials (D 790), using a sup-

port length-to-thickness ratio of 40:1. Specimens were of 8.9- by 2.54-cm (3.5- by 1-in.) configuration. All mechanical tests were performed on fifteen samples.

Moisture absorption behavior of graphite/composites was determined by exposing test specimens to 95% relative humidity at 71°C (160°F). Tests were conducted on [0/90]_s 76-mm (3.0-in.) square specimens of 1.0-mm (0.04-in.) thickness to obtain the rate of moisture absorption and the maximum moisture absorbed by weight. Specimens were dried before testing at 77°C (170°F) in vacuum until their weights remained unchanged. The tests were performed on three samples. Coefficient of thermal expansions was measured over the temperature range of -184 and +100°C (-184 and +212°F) using a dilatometer. The outgassing/volatiles testing was conducted according to the ASTM Test for Total Mass Loss and Collected Volatile Condensable Materials from Outgassing in a Vacuum Environment (E 595).

Results and Discussion

Mechanical Characterization

A statistical room temperature design allowables property database for use in the structural analysis of primary and secondary components was generated from unidirectional IM7/PEEK laminates fabricated from prepreg tapes of three different production lots. Average property values as well as the summary statistics of the mechanical property testing are summarized in Table 1. Typical fractures of the 0° and 90° tension specimens, 0° and 90° compression specimens, and ±45° tension specimens are shown in Figs. 3 through 7, respectively. All specimens produced acceptable failure modes consistent with the specific property of the composite.

The 0° tensile, compressive, and flexural moduli were statistically equivalent. As with most graphite fiber composites, the 0° compressive strength was lower than the tensile strength. The 0° flexural strength was between the tensile and compressive values, as would be expected. The in-plane shear strength and modulus obtained from the tension tests on [±45] laminates were 160 and 4.9 GPa (23.3 ksi and 0.71 psi). The high in-plane shear strength is due to the large axial failure strain, typically in the range from 15 to 20%. This is attributed to the high ductility of the polymeric matrix and a strong fiber-matrix interfacial adhesion. The linear coefficients of thermal expansion at 0° and 90° in a unidirectional layup are summarized in Table 1.

Figure 8 presents a comparison of the IM7/PEEK mechanical properties with the corresponding IM7/8551-7 epoxy properties [14]. The 0° and 90° tensile properties of the IM7/PEEK composite are approximately the same as the IM7/8551-7 epoxy values. The compressive strength of the IM7/PEEK is slightly lower than that of the IM7/8551-7 epoxy. The IM7/PEEK composite has a higher shear strength but a lower shear modulus compared to those property values measured for the IM7/8551-7 epoxy.

A processing study was conducted to determine the effect of varying consolidation temperatures on laminate properties. Three laminates were consolidated using the superplastic diaphragm-forming process at three temperatures: 371, 388, and 404°C (700, 730, and 760°F). Flexural properties were determined on specimens taken from the laminates; results are summarized in Table 2. The overall results for these study laminates indicate that varying processing temperatures within the range of 371 to 404°C (700 to 760°F) does not affect the basic mechanical properties.

Fiber/Resin Adhesion

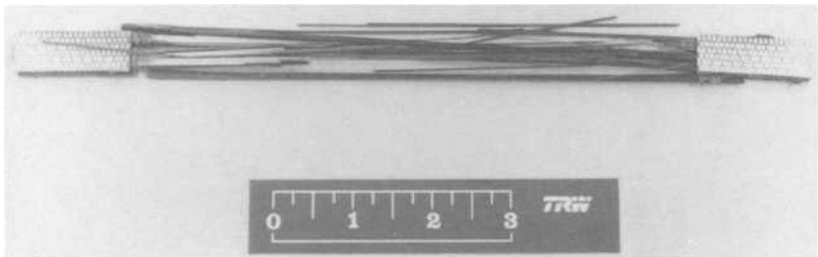
The fiber-matrix interfaces of the PEEK thermoplastic composites were examined by SEM to obtain a better understanding of fiber/resin adhesion. Figure 9 shows photomicrographs of the transverse tensile fracture surfaces of the specimens fabricated from the unidirectional

TABLE 1—*Properties of IM7 graphite/thermoplastic PEEK.*

Property		Units	$\bar{X}$	s^a	n^b
Fiber Volume	Vf	%	62.3	0.4	15
Modulus					
0-deg tensile	E-11t	GPa (Msi)	153 (22.1)	7.7 (1.1)	15
0-deg compressive	E-11c	GPa (Msi)	153 (22.2)	4.6 (0.7)	15
90-deg tensile	E-22t	GPa (Msi)	9.1 (1.31)	0.6 (0.09)	15
90-deg compressive	E-22c	GPa (Msi)	8.8 (1.28)	0.9 (0.12)	15
Strength					
0-deg tensile	S-11t	MPa (ksi)	2500 (362.4)	260 (37.8)	15
0-deg compressive	S-11c	MPa (ksi)	908 (131.7)	70 (10.1)	15
90-deg tensile	S-22t	MPa (ksi)	84 (12.2)	7.5 (1.1)	15
90-deg compressive	S-22c	MPa (ksi)	168 (24.3)	10.8 (1.6)	15
Strain at Failure					
0-deg tensile	e-11t	%	1.52	0.12	15
0-deg compressive	e-11c	%	0.64	0.10	15
90-deg tensile	e-22t	%	1.03	0.10	15
90-deg compressive	e-22c	%	1.80	0.25	15
In-Plane Shear Properties					
Modulus	G ₁₂	GPa (Msi)	4.9 (0.71)	0.7 (0.1)	15
Strength	τ_{12}	Mpa (ksi)	160 (23.3)	49 (7.2)	15
Flexural Properties					
Modulus		GPa (Msi)	153 (22.1)	7.7 (1.1)	15
Strength		Mpa (ksi)	2193 (318.0)	168 (24.4)	15
Poisson's Ratio	ν_{12}		0.33	0.03	15
Coefficient of Thermal Expansion	α_{11}	10 ⁻⁶ in/in - °C	-0.07 (-0.04)	0.05 (0.03)	3
	α_{22}	(10 ⁻⁶ in/in - °F)	25.7 (14.3)	0.74 (0.41)	3

a) s = Standard deviation
b) n = Number of specimens

graphite/PEEK prepreg. As can be seen from the photomicrographs, good wetting of the graphite fibers and excellent adhesion were observed for the PEEK matrix resin. There is considerable plastic deformation and almost total cohesive failure with almost all fibers coated by the PEEK matrix. These findings are consistent with the corresponding high transverse tensile properties measured for the unidirectional prepreg composites (see Table 1).



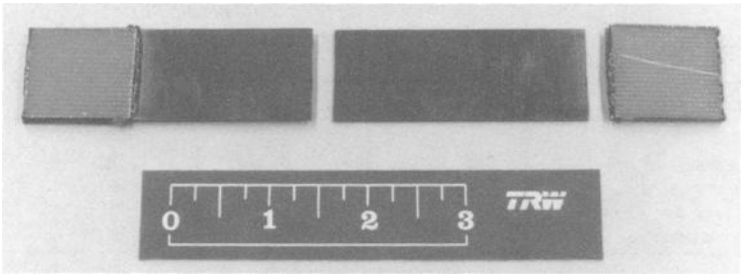


FIG. 4—Typical fracture of 90° tension specimen.

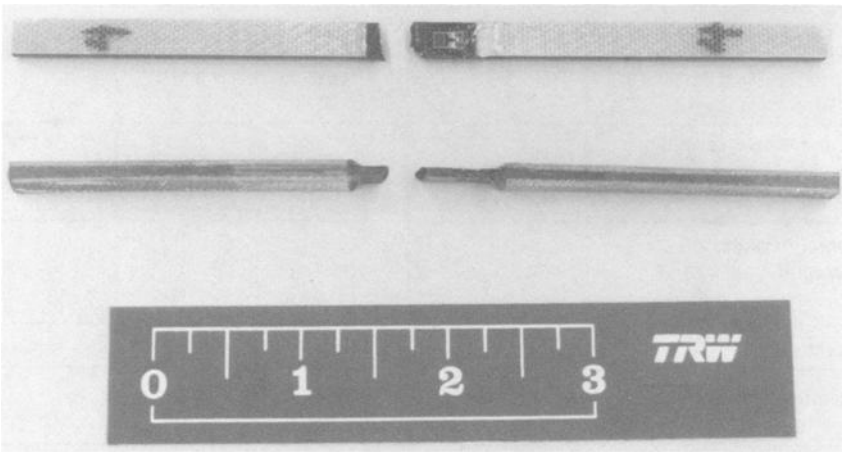


FIG. 5—Typical fracture of 0° compression specimen.

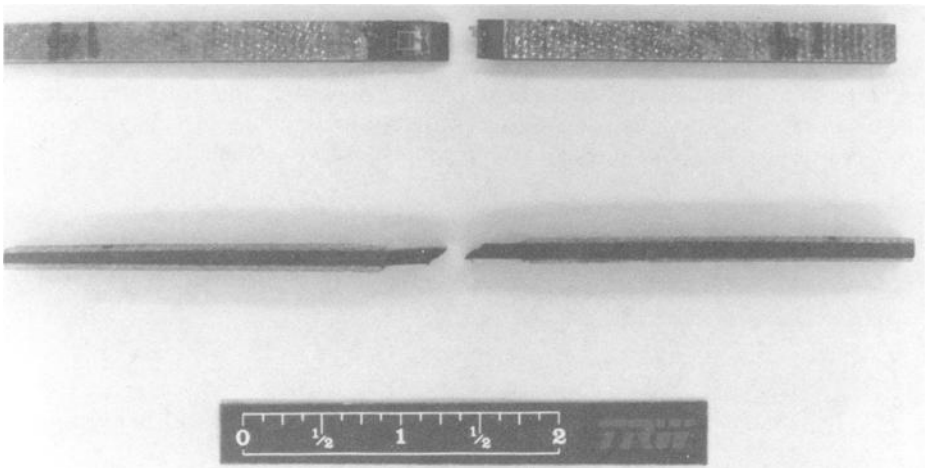


FIG. 6—Typical fracture of 90° compression specimen.

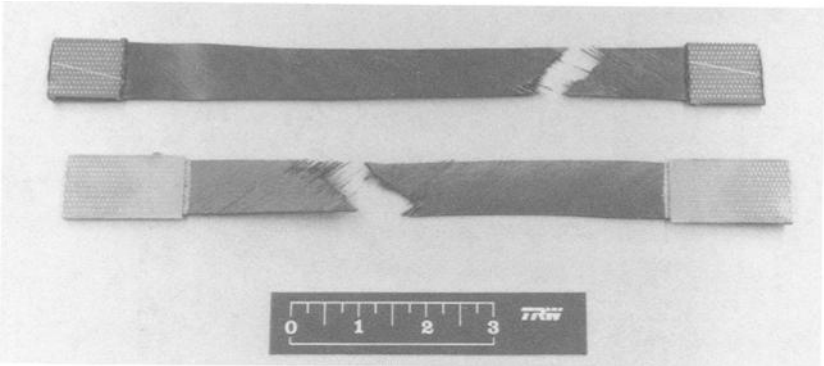


FIG. 7—Typical fracture of $\pm 45^\circ$ tension specimen.

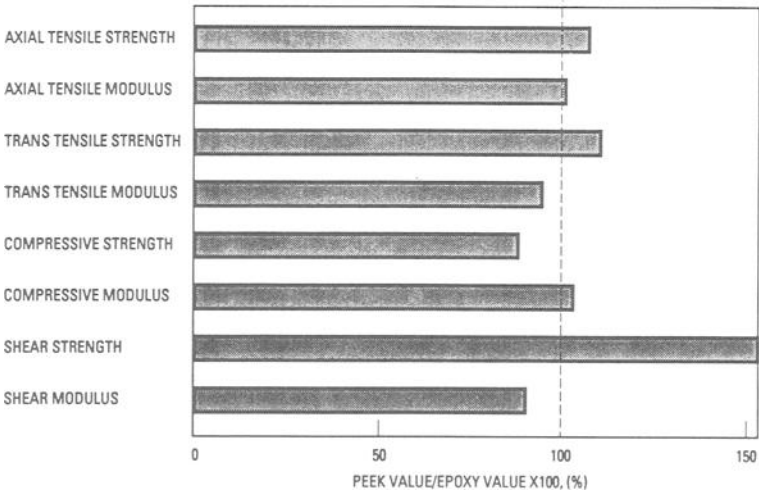


FIG. 8—Comparative properties of IM7/PEEK and IM7/8551-7 epoxy laminates [14].

TABLE 2—Effect of processing temperatures on the flexural properties of IM7/PEEK.

Processing Temperature, °C (°F)	Flexural Properties				<i>n</i> ^b
	Strength	SD ^a	Modulus	SD	
	MPa (ksi)	MPa (ksi)	GPa (Msi)	GPa (Msi)	
371°C (700°F)	2272 (329.4)	119 (17.2)	156 (22.6)	5.4 (0.78)	8
388°C (730°F)	2115 (306.6)	155 (22.5)	148 (21.4)	3.5 (0.50)	8
404°C (760°F)	2366 (343.0)	92 (13.3)	165 (24.0)	5.2 (0.76)	8

^a SD = standard deviation.

^b *n* = number of specimens.

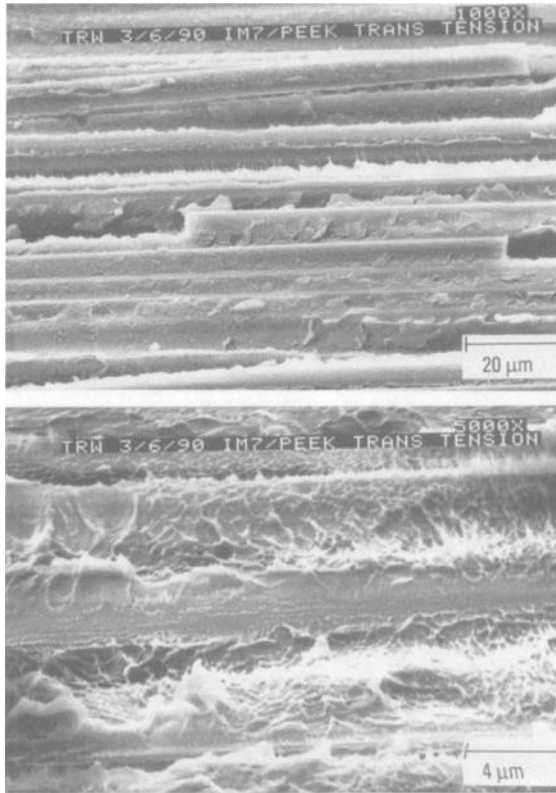


FIG. 9—Scanning electron micrographs of transverse tensile fracture surfaces of specimens prepared from graphite/PEEK prepreg tapes.

The strong interfacial adhesion observed for the PEEK specimens has been previously attributed to PEEK crystallization on the graphite fibers, which makes a strong interfacial bond as indicated by the results of the transverse tension testing (see Table 1). In their study on the crystallization of PEEK in graphite fiber composites, Lee and Porter [15] observed that destroying nuclei in the PEEK matrix by long preheating enhanced crystallization on the graphite fibers rather than in the PEEK matrix during cooling. When the time at which the polymer was held at the melt temperature before cooling was increased, greater polymeric matrix adhesion to the graphite fibers leading to higher transverse tensile strength was reported. More recent studies [1,3,4,16] have shown morphological features suggesting that the graphite surface acts as nucleation sites for the PEEK resin.

Outgassing

Graphite/epoxy is hygroscopic and, depending on the particular epoxy, can absorb between 0.80 and 1.30% moisture by weight. In the vacuum of space, graphite/epoxy desorbs its absorbed moisture, which causes problems in a cryogenically cooled infrared sensor system in which graphite/epoxy is used as a structural material. The desorbed moisture condenses on the cryo-cooled mirror surfaces and causes image degradation. Outgassed secondary species

The outgassing/volatiles testing indicated that the graphite/thermoplastic PEEK had total mass loss (TML) values of less than 0.053% and collected volatile condensable materials (CVCM) of less than 0.0004%. In comparison, TML and CVCM values for a IM7/8551-7 epoxy were measured as 0.232 and 0.0091%, respectively. Values below 1.0% TML and 0.1% CVCM have been acceptable for current spacecraft performance needs, but the requirements are expected to become more stringent for future surveillance spacecraft systems.

Water Absorption Study

Stringent dimensional stability is required for high-performance space science instruments and optical benches. A major problem with the use of epoxy-matrix components for dimensionally stable structures is that epoxy absorbs and desorbs moisture in a vacuum, which can cause large dimensional changes. The loss of this moisture in space is accompanied by a dimensional change as reflected by the coefficient of moisture expansion (CME). Many methods have been used to circumvent this outgassing problem, the most effective of which has been the use of organic or inorganic barrier coatings or both for sorption/desorption control. Another method involves time-consuming degassing and isolation procedures to prevent graphite/epoxy from absorbing moisture. Both of these techniques have associated difficulties that are expensive and time-consuming.

Maximum Water Absorption and Diffusion Coefficient—The effect of moisture exposure on graphite/PEEK and graphite/8551-7 epoxy composites was examined. Figures 10 and 11 show the moisture absorption behavior for IM7/PEEK and IM7/8551-7 epoxy laminates. Both materials approached saturation after approximately 100 hours of exposure to a humid environment. The PEEK laminate exhibited a very low saturation value of 0.18%, whereas the epoxy composite absorbed as much as 0.8% by weight of water.

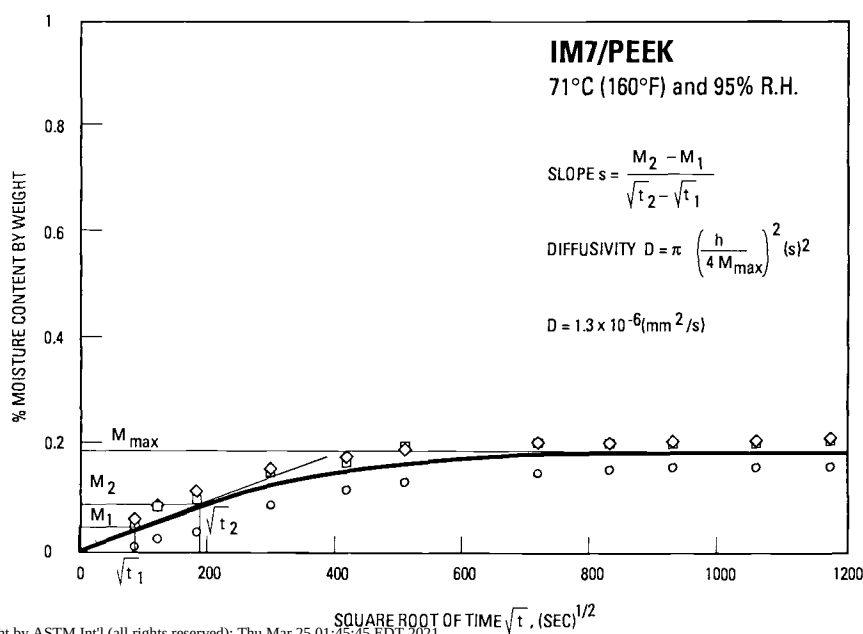


FIG. 10—Moisture absorption behavior of IM7/PEEK laminate.

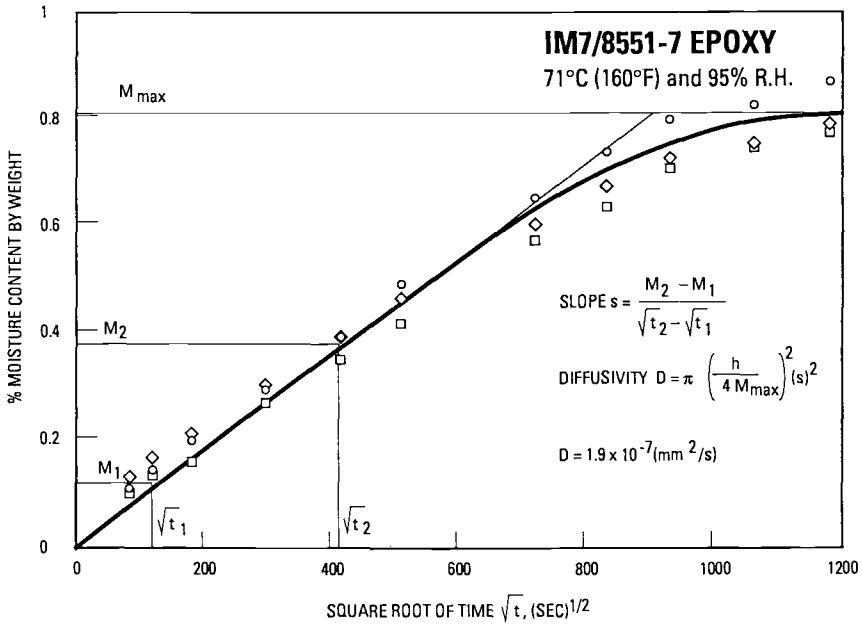


FIG. 11—Moisture absorption behavior of IM7/8551-7 epoxy laminate.

The diffusivity coefficient was determined by the following expression:

$$D = \pi \left(\frac{h}{4M_{\max}} \right)^2 \frac{M_2 - M_1}{\sqrt{t_2} - \sqrt{t_1}}$$

where

D = the diffusivity coefficient in mm^2/s

h = the laminate thickness in millimetres

$M_{\max}$ = the maximum moisture content in percentage for a given environment, and

M_1 and M_2 = percent moisture contents in the laminate after t_1 and t_2 s of exposure to the given environment.

The IM7/PEEK composite was determined to have a diffusivity coefficient of $1.3 \times 10^{-6} \text{ mm}^2/\text{s}$, which is approximately ten times the diffusivity coefficient of $1.9 \times 10^{-7} \text{ mm}^2/\text{s}$ measured for the IM7/8551-7 epoxy composite. Therefore, IM7/PEEK absorbs moisture much faster than IM7/8551-7 epoxy.

Coefficient of Moisture Expansion—The coefficient of moisture expansion (CME) is defined as a change in length per unit length per weight percent of water absorbed at constant temperature and pressure. Dimensional changes caused by the loss of water in space can be either negative or positive in a multidirectional angle-ply laminate. Prediction of CME in various in-plane and thickness directions requires a knowledge of axial (fiber-direction) and transverse coefficients of moisture expansion for unidirectional laminates. CME values are readily incorporated into finite element and other computer codes of laminate design. The CME values for IM7/PEEK were determined by Composite Optics, Inc., San Diego, using their optical comparator. The measurement of the coupon strain (change in length per unit length) is made by an optical technique that determines the difference between the coupon strain and the strain

TABLE 3—Coefficient of moisture expansion test results for IM7/PEEK laminates

Material	Fiber Direction	Specimen Number	Moisture Content, %M	Strain, $\times 10^{-6}$ in/in	CME, ^a $\times 10^{-6}$ in/in/%M
IM7/PEEK	axial	1	0.11		
		2	0.11	8	76
		3	0.11	8	76
	transverse	1	0.12	312	2602
		2	0.11	320	2913
		3	0.11	338	3071
T300/934 ^b	axial	1	1.22	25	21
	transverse	1	1.28	2435	1902

^a CME = Coefficient of moisture expansion^b Data from NASA ACTS (Advanced Communication Technology Satellite) Study

of a standard material. The standard material is ultra-low expansion-fused silica. The standard is calibrated for CME and, when the absolute strain of the standard is added to the measured differential strain between the coupon and the standard, the absolute strain of the coupon is the result. All specimens were preconditioned in vacuum for 5 days at 63°C (145°F). Strain measurements and percent water absorbed by weight were made on specimens exposed to 90% relative humidity and 37°C (99°F) for 31 days.

Table 3 summarizes the results for three IM7/PEEK [0]_z specimens: percent moisture present in the sample after each exposure, the measured strain, and the calculated CME at saturation. The data for the three specimens are very consistent in terms of both moisture content and strain measurements. For example, the final moisture content of the three transverse specimens at saturation was 0.12, 0.11, and 0.11%, respectively. Similarly, the final strain readings for the three samples were +312, +320, and +338 μ in/in, respectively.

For comparison, literature values of moisture absorption for typical graphite/epoxy laminates are also presented in Table 3. Compared to T300 graphite/934 epoxy (data not available for IM7 graphite/8551-7 epoxy), the IM7 graphite/PEEK displayed lower equilibrium saturation moisture uptake and lower strain measurements in both the axial and transverse directions. However, the calculated CME values for the graphite/PEEK were higher than for graphite/epoxy, which is attributed to the division of the strain value by a fractional moisture content value.

Conclusions

- 1 Composite laminates fabricated from the unidirectional PEEK prepreg tapes exhibited very respectable mechanical properties that will meet future spacecraft requirements.
- 2 PEEK shows excellent potential as a polymeric matrix for spacecraft structural components. The graphite/PEEK composite exhibited low moisture absorption and acceptable outgassing properties.
- 3 The IM7/PEEK matrix laminates absorb moisture much faster than the IM7/8551-7 epoxy matrix laminates. But the equilibrium moisture content of the PEEK matrix laminates exposed to 95% relative humidity is significantly lower than that of the epoxy matrix laminates, 0.18% versus 0.80%, respectively.
- 4 Thermoplastic PEEK composite laminates fabricated by superplastic diaphragm-forming methods exhibit excellent fiber/matrix distribution and interfacial fiber/resin adhesion.

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Hot/Wet Testing of Celion 3000/PMR-15 Coupon Specimens

REFERENCE. Kinman, E. B., "Hot/Wet Testing of Celion 3000/PMR-15 Coupon Specimens," *Composite Materials: Testing and Design (Tenth Volume)*, ASTM STP 1120, Glenn C. Grimes, Ed., American Society for Testing and Materials, Philadelphia, 1992, pp. 131-141.

ABSTRACT. Mechanical properties testing of a graphite/polyimide (Celion 3000/PMR-15) fabric was performed in primarily four environments — 54°C (–65°F) as fabricated, room temperature as fabricated, 204°C (400°F) after moisture conditioning, and 316°C (600°F) after moisture conditioning. Specimens that had been moisture conditioned before testing were considered "wet." To accomplish this hot/wet testing, special strain-gaging techniques were developed to improve strain measurement in the severe environments. Heating techniques were also developed to heat the specimens quickly to retain sufficient moisture at the time of failure. This test program demonstrated that moisture, temperature, and specimen geometry can each dramatically affect test results. The program also demonstrated the need for industry standards for high-temperature testing and moisture level requirements. Without such standards, data thought to be hot/wet test data may actually be hot/dry test data. Such a discrepancy appears to have significant ramifications according to this program's test results.

KEYWORDS. composite materials, graphite-polyimide composites, mechanical properties, moisture content, high temperature tests, strain gages

New high-temperature composite materials are constantly being developed for applications in present and future high-performance aircraft. But, before meaningful design property allowables can be developed, a particular material's mechanical properties must be ascertained during exposure to environments simulating real aircraft conditions. Obtaining these data is often difficult, however, because of the severe environmental conditions that specimen instrumentation must endure.

A two-year specimen test program was undertaken to generate data and design allowables for material that consisted of Celion 3000 (C3000) carbon fiber tows woven into an eight-harness satin weave fabric and impregnated with PMR-15 polyimide resin. The C3000 carbon fiber tows were made from a polyacrylonitrile (PAN) precursor. Allowables were to include data at 204 and 316°C (400 and 600°F) after moisture conditioning (wet). Specimens tested as fabricated were considered "dry," although their actual moisture content was 0.1 to 0.4% by weight. Travelers representing wet specimens typically had a moisture content of 1.21 to 1.33% by weight. Ply properties were evaluated and then used to predict laminate behavior under load. The properties of the laminate were then verified with tests. Table 1 shows the test matrix for the C3000/PMR-15 fabric lamina tests. Table 2 presents the test matrix for the C3000/PMR-15 fabric quasi-isotropic laminate tests and is included for information only.

Before the fabrication of test specimens, test panels had to pass a fracture critical nonde-

¹ Senior materials and processes engineer, LTV Aircraft Products Group, P.O. Box 655907, M/S 194-51, Dallas, TX 75265-5907.

TABLE 1—Test matrix for C3000/PMR-15 fabric lamina test specimens

TEST TYPE		TEST ENVIRONMENT				TOTAL TESTS
		– 54°C/DRY	RTD	204°C/WET	316°C/WET	
TENSION	0°	18	18	18	23	77
	90°	18	18	20	23	79
COMPRESSION	0°	19	18	20	25	82
	90°	18	18	21	25	82
IN-PLANE SHEAR (±45° TENSION)		18	18	20	23	79
TOTAL TESTS						399

structive inspection (NDI) C-scan to verify the quality of the test panels For the C3000/PMR-15 lamina tests, three batches of material were used with a minimum of six tests per batch for each test type/environment The exceptions were additional tests that were performed to verify data from previously tested specimen sets and additional compression specimens that were used to investigate hot/dry compressive properties and determine the effect of specimen geometry on lamina test values Table 3 gives typical results of the acceptance tests for the five batches of material Figure 1 shows the two compression specimen configurations Table 4 presents the test matrix for the additional compression tests

TABLE 2—Test matrix for C3000/PMR-15 fabric laminate test specimens

TEST TYPE		TEST ENVIRONMENT			TOTAL TESTS
		RT/DRY	204 C/WET	316 C/WET	
UNNOTCHED	TENSION	6	—	—	6
	COMPRESSION	6	—	—	6
0 64-cm OPEN HOLE	TENSION	6	5	4	15
	COMPRESSION	6	3 ^a	4 ^a	13
BEARING	SINGLE SHEAR	8	—	—	8
	DOUBLE SHEAR	7	3	4	14
IMPACT DAMAGE TOLERANCE – COMPRESSION ^b		17	—	—	17
BASELINE COMPRESSION FOR IMPACT ^c DAMAGE TOLERANCE		3	—	—	3
BOLT BEARING BYPASS – TENSION ^d		23	—	—	23
INTERLAMINAR SHEAR NOTCHED COMPRESSION		6	6 ^a	5 ^a	17
FLATWISE TENSION (RECTANGULAR)		6	5 ^a	5 ^a	16
FLATWISE TENSION ^d (1 90 cm ROUND)		5			5
TOTAL TESTS					143

^aTHESE SPECIMENS WERE NOT MOISTURE CONDITIONED

^bSPECIMENS IMPACTED AND THEN COMPRESSION TESTED

^cSPECIMENS NOT IMPACTED BUT COMPRESSION TESTED TO DETERMINE BASELINE PROPERTIES

^dSPECIMENS TESTED IN TENSION AND BEARING SIMULTANEOUSLY SEVERAL TENSION/BEARING RATIOS WERE EVALUATED

TABLE 3—Results of LTV acceptance tests for five batches of C3000/PMR-15 fabric prepreg.

	LTV Specification Requirement ^a	Typical per LTV Tests ^b
Resin content, percent by weight	35 ± 3	35.8
Density, g/cm ³	1.57 ± 0.09	1.56
Fiber volume, %	62 ± 3	63.7
Resin volume, %	38 ± 3	33.6
Void volume, %	3 maximum	2.7

^a LTV’s preliminary specification.
^b For the five batches of material used for these tests.

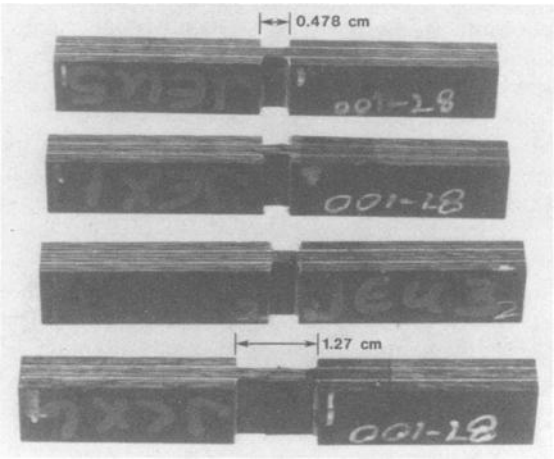


FIG. 1—Two compression specimen configurations.

TABLE 4—Test matrix for additional C3000/PMR-15 fabric compression lamina test specimens.

TEST TYPE	TEST ENVIRONMENT									TOTAL TESTS
	−54°C	RT		204°C		260°C		316°C		
	DRY	WET	DRY	WET	DRY	WET	DRY	WET	DRY	
0 ^{oa}	18	7	18	20	6	19	6	6	5	105
0 ^{ob}	—	—	—	—	—	—	—	—	7	7
90 ^{oa}	—	—	—	18	—	18	—	—	—	36
									TOTAL TESTS	148

^aSPECIMENS WITH 0.478-cm TEST SECTION
^bSPECIMENS WITH 1.27-cm TEST SECTION

Procedures

Installation of Strain Gages

All hot/wet specimens were strain gaged before moisture conditioning to avoid drying as a result of the temperature of the strain gage adhesive cure cycle Table 5 presents the strain gages, adhesives, and adhesive cure parameters used for this test program

The gages used during this test program were chosen on the basis of grid alloy, encapsulation material, grid configuration, and lead wire attachment configuration The nickel-chromium grid alloy, similar to Karma, was needed for the high test temperatures The encapsulation films (glass-fiber-reinforced epoxy-phenolic and polyimide) were needed to aid in protecting the gages from corrosion during moisture conditioning and handling These encapsulation materials were also suitable for use at their respective test temperatures, 204 and 316°C Kap-ton® polyimide film was used under gage whiskers to prevent them from becoming embedded in excess adhesive during the vacuum-bagging cure cycle The various grid and lead attachment configurations chosen for each specimen geometry were important in improving the ease of strain gage application and wiring, as well as in obtaining strain values in the desired direction These configurations enhanced the durability of the strain gage installations during routine handling Gage lots were recorded as the gages were installed, and their gage lot data sheets were retained to calculate high-temperature gage factors as needed

Stranded, Teflon®-coated 32-AWG wire was chosen as the lead wire because it provided a good connection to gage whiskers, flexibility, and withstood the high test temperatures This flexibility was significant because it increased the durability of the gage installations during handling

Vacuum bagging was selected to apply pressure during the adhesive cure cycle of the strain gages to be used at 316°C (600°F) This method was chosen because of the results of a previous LTV Aircraft Products Group high-temperature gaging investigation The pressure applied to the gages was multiplied by symmetrically placing pieces of aluminum sheet that were greater in area on top of silicone rubber pads used over the strain gages. Steel weights were also used to increase pressure by placing the weights above the padded aluminum sheets, but over the

TABLE 5—Strain gages and adhesive cure cycle parameters for hot/wet testing

STRAIN GAGES	TEST TEMPERATURE	
	204°C	316°C
	MICRO-MEASUREMENTS WK TYPE (OPTION W)	MICRO-MEASUREMENTS WK TYPE
ADHESIVE	M-BOND 610 (MICRO-MEASUREMENTS)	PLD-700 (BLH ELECTRONICS)
ADHESIVE CURE		
TEMPERATURE	177°C (350°F)	260°C (500°F)
HOLD TIME	1 HOUR	1 HOUR
PRESSURE	0 207 – 0 276 MPa (30 – 40 PSI)	0 207 – 0 276 MPa (30 – 40 PSI)
APPROX TIME IN OVEN	5 HOURS + COOL DOWN	6 HOURS + COOL DOWN
POST CURE		
TEMPERATURE	232°C (450°F)	260°C (500°F)
HOLD TIME	2 HOURS	2 HOURS
APPROX TIME IN OVEN	4 HOURS + COOL DOWN	4 HOURS + COOL DOWN

bagging material. Postcures were used for high-temperature gage installations to improve the test performance of the strain gage adhesive and were performed without pressure applied. After exposure to cure and postcure temperatures, the specimens were allowed to cool to near ambient temperature while remaining in the closed oven. This slow cool down was necessary to prevent microcracking of the specimens.

After bonding strain gages, proper gage positioning was verified. The vacuum-bagging procedure sometimes caused movement of the gages. The specimens with an unacceptable gage orientation were gaged again. Once the gage position was acceptable, the Teflon-coated lead wires were attached to the gage whiskers. Most gages used at 316°C were wired with silver solder using a small oxyacetylene torch. Some gage whiskers were attached to the lead wires using a spot-welding technique. These welds appeared to perform as well or better than the silver solder joints, but the welding procedure required a great deal of practice for consistency and, if performed incorrectly, the gage whiskers would often be too short to attempt welding again. Therefore, regaging would become necessary, resulting in significantly increased test costs. For these reasons, the silver solder and its simple soldering procedure were most often used for the 315°C gage installations.

The gages and lead attachments used at 204°C (400°F) were coated with M-Bond 610 adhesive to retard the corrosive affects of moisture conditioning. The adhesive was then cured, but not postcured. Gage installations used at 316°C were not coated before moisture conditioning. Gage coatings used on trial specimens seemed to cause premature failure of the gage during testing. The gage coatings also prevented lead wire attachment repairs, if needed, before testing. High-temperature tapes and adhesives in conjunction with room temperature vulcanization (RTV) silicone rubber were the treatments evaluated for this purpose. The adhesive coating appeared to degrade the gage performance as a result of thermal stresses. This conclusion was supported by inspections of the encapsulation area on failed gages. It was found that most 316°C gage installations left uncoated during moisture conditioning performed better than coated installations, if the specimens were removed from the moisture chamber within a few days after reaching the required moisture content. The use of Mylar® and Teflon shrink tubing on the whisker solder joints to prevent possible shorts from crossed whiskers was also investigated. Both were usable, but the Mylar did not permit lead wire attachment repairs. Also note that a group of gages that were wired using the spot-welding method in conjunction with shrink tubing performed better than other gages on similar specimens, but this was not determined until after all the strain gages were installed.

Moisture Conditioning

Specimens that required moisture conditioning before testing were accompanied by traveler specimens throughout humidity exposure and elevated temperature testing. These travelers were made from the same panels as the test specimens they represented and were cut into 2.54-by-7.62-cm (1- by 3-in.) portions. The moisture content of the specimens was based on the weight changes of the representative travelers. Specimens in groups of six and their travelers were moisture conditioned and tested as a set. The specimens and their travelers were first dried to a constant weight in an air-recirculating oven at 77°C. After drying, the specimens and their travelers were exposed to 95 to 100% relative humidity at 71°C until the travelers of the set reached a weight gain of 1.1 to 1.4%. This weight gain represents a "worst case" service condition expected for the material's application.

After a specimen and traveler set attained a moisture content above 1.1% by weight, the set was sealed in a moisture proof (MIL-B-131) aluminum foil bag until testing. If the specimens were not tested within five days, they were returned to the humidity chamber to await rebagging and then testing. Traveler weight measurements were recorded weekly during moisture

conditioning and on the day specimens were removed from the humidity chamber Travelers used for moisture verification were also weighed immediately before and after testing

Heating of Test Specimens

During a preliminary test procedures investigation, it was determined that the environmental test chambers typically used to heat specimens caused undesired effects on testing at the 204 and 316°C test temperatures These chambers enclosed the grips or the compression test fixtures during testing This presented several problems including ones of grip expansion and extended heating times as a result of increased heat absorption and conduction by and through the metal parts Heat conducting through the load train also increaesd the chances of over-heating the load cell, even though a cooling coil was in place to prevent such an occurrence The 316°C test temperature could also have degraded the condition of grips and fixtures during the many tests that were to be performed

It was decided to heat the test section of the specimens, but not the tension testing grips or the compressive loading ram A blower, a heater, and a temperature controller were the main components of the hot-air equipment used in this test Two insulated stainless steel boxes that enclosed the specimen's test section were fabricated, one for tension testing and one for compression To provide easy access to the inside of the boxes, each box was comprised of two sections, one of which slipped over the other A thermocouple was mounted at the inlet of the box to provide immediate feedback for the heat-gun temperature controller, and one was also attached to the specimen to allow monitoring of its temperature Figure 2 presents a schematic of the heating setup and Fig 3 shows the test setup for 204° to 316°C compression testing The heater had a fast response time which also contributed to reducing the time needed to reach the test temperatures

It was important to heat the specimens quickly for several reasons It had been determined that graphite/PMR-15 used in propulsion-related structural applications could be exposed to very rapid heat rise rates, possibly after moisture saturation Therefore, our goal was to obtain test data from specimens that had been moisture conditioned and retained a significant amount of moisture, approximately half of the saturation moisture content, at the time of failure Travelers were used to determine the moisture content of the test sections at failure for the different test types and environments Several specimens for a given type and environment were heated and tested Then a representative traveler would be exposed to the same air temperature and average time parameters This was usually performed with multiple specimen sets for each test type and environment In the case of in-plane shear specimens tested at 316°C,

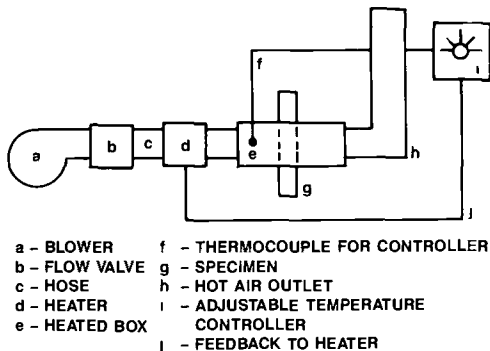


FIG. 2—Heating setup for 204 to 316°C testing

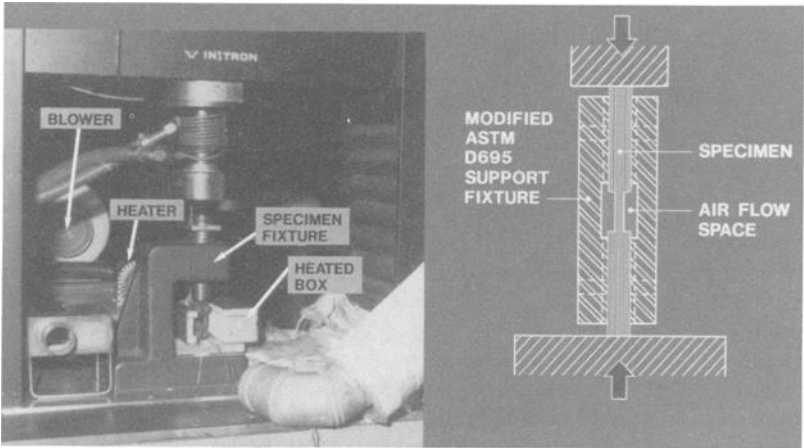


FIG. 3—Test setup for 204 to 316°C compression testing.

the loading rate was increased to shorten the high-temperature exposure after two specimens and a representative traveler showed that the specimens were practically dry at the time of failure. While the increased loading rate added another variable to the 316°C in-plane shear test data, it was determined that obtaining data from relatively “wet” specimens was critical for our application. The remaining specimens tested with the increased loading rate had a much higher moisture content at the time of failure, as seen in Table 6.

While it was important to heat the specimens quickly, it was also important not to shock the specimen or the gages thermally. Because the heat gun was capable of reaching the set temperature of the controller within seconds, it was necessary to establish a heat ramp. A soak time was also established to give the specimen sufficient time at the test temperature to reach a thermal steady-state condition. To determine the necessary soak time, trial specimens with thermocouples attached at several positions within their heated test sections were placed in the

TABLE 6—Average moisture content of traveler specimens.

TYPE TEST SPECIMEN	MOISTURE CONTENT AFTER CONDITIONING (%)	MOISTURE CONTENT AT FAILURE (%)
AT 204°C/WET		
0° TENSION	1.28	0.77
90° TENSION	1.28	0.77
0° COMPRESSION	1.33	0.89
90° COMPRESSION	1.33	0.89
±45° TENSION (SHEAR)	1.28	0.62
AT 316°C/WET		
0° TENSION	1.21	0.33
90° TENSION	1.24	0.17
0° COMPRESSION	1.26	0.57
90° COMPRESSION	1.26	0.57
±45° TENSION (SHEAR) ^a	1.28	0.44

^a APPLICABLE FOR SPECIMENS TESTED WITH A CROSSHEAD SPEED OF 0.81 cm/min

hot-air box and heated to the elevated test temperatures. It was found that all the trial specimens were at a steady-state condition by 15 min. Therefore, our standard soak time was 15 to 2 min.

General Hot/Wet Test Procedures

Travelers used to measure moisture content after testing were weighed immediately before testing. All gage installations were checked before humidity exposure, yet after humidity exposure, the installations often had high resistances which were traced to the area between the gage grid and the gage whisker or solder dot. Adjustments were then made to the installation or the instrumentation as required. The specimen was installed into the test setup and preloaded to approximately 1% of expected failure load. Then the specimen's alignment within the fixture was checked using the strain measurements from the back-to-back axial gages on the specimen. Alignment was considered acceptable if these strains were within 5% of one another. If not, several attempts would be made to improve the alignment by adjusting the specimen's position within the test fixture. If the alignment measurement did not improve, it was considered an anomaly of the specimen. Then the specimen was placed in its most aligned position, the strain measurements were recorded, and the test proceeded. Next, the specimen was heated and soaked at the test temperature as previously described. The ASTM Test for Tensile Properties of Fiber-Resin Composites (D 3039) procedures were followed for the tension-loaded lamina test specimens. However, the 316°C in-plane shear specimens were tested at a faster than specified strain rate of 8.2×10^{-4} m/m-s (0.049 in./in.-min) to retain moisture until the time of failure. The ASTM Test for Compressive Properties of Rigid Plastics (D 695) procedures were followed for the compression lamina test specimens, however, modified specimens and a modified support jig were used as shown in Figs. 1 and 3.

Results and Discussion

Effect of Specimen Geometry

As mentioned previously, two configurations of compression specimens were tested. Both were modified ASTM Test D 695 specimens with straight sides and tabs, but one had a 127-cm test section and the other had a 0.478-cm test section. The 127-cm test section specimen was originally chosen to obtain both strength and strain data from each specimen. The 127-cm test section allowed the installation of back-to-back strain gages, while the 0.478-cm test section would not. However, when the hot/wet test data from the 127-cm test section specimens were lower than anticipated, it was thought that the longer unsupported test section could have caused premature failures. The 0.478-cm test section specimens were then fabricated and tested. The failure strengths of the 0.478-cm test section specimens were higher than those of the first configuration as seen in Table 7.

The test results from the two configurations of compression specimens demonstrated that

TABLE 7—*Effect of test section length on C3000/PMR-15 fabric
0° compressive strength*

Test Environment	Percent Strength Increase, 0.478-cm Test Section over 127-cm Test Section
Room temperature/dry	21
204°C/wet	23
316°C/wet	12

TABLE 8—*Effect of moisture on elevated temperature C3000/PMR-15 fabric 0° compressive strength*

	204°C/Wet	316°C/Wet
Moisture content, % (based on traveler specimens)		
● after conditioning	1 33	1 26
● at time of failure	0 89	0 57
Hot/wet strength, percent of hot/dry strength	65	36

strength test values can vary significantly with specimen design and geometry. After reviewing all the test results, typical and design allowables, and the analytical correlation between the lamina and laminate test results, it was decided to use the 1 27-cm test section compression test results for design allowables. The results from these tests provided the best correlation with the quasi-isotropic test results. Analytical predictions from the quasi-isotropic specimen test values indicated that the modified ASTM Test D 695 specimen with a 0 478-cm test section gave high strength values.

Compression Testing

When testing fabric laminates, use the modified ASTM Test D 695 specimen with the 1 27-cm test section after checking for column stability. It is also recommended that separate low- and high-temperature compression testing standards be developed to account for different fixture considerations. Thermal expansion, thermal conductivity, and thermal mass are all issues because of the great amount of metallic fixturing typically used. Icing of close tolerance fixturing is a problem at low temperatures, and instrumentation is frequently restricted by the limited access caused by the fixturing.

Effects of Absorbed Moisture and Temperature

Since additional compression testing was performed, hot/dry and hot/wet 0° compression data were available for comparison. Table 8 presents the moisture content of the wet specimens and the wet strength as a percentage of the dry strength for tests performed at 204 and 316°C.

For all specimen types, as the test temperature increased above room temperature, the strength values decreased. However, in most cases moisture conditioning was an added variable. The additional 0° compression tests also permitted us to compare room temperature/dry and hot/dry strength values. Table 9 gives the results of the dry and wet compression tests. Table 10 presents the results of hot/wet tension, compression, and in-plane shear tests.

TABLE 9—*Typical 0° compressive strength versus temperature*

Specimen Type	Percent of RTD Strength ^a			Coefficient of Variance		
	204°C	260°C	316°C	204°C	260°C	316°C
0° compression (1 27-cm test section)						
dry			48			11 0
wet	58		17	12 0		20 0
0° Compression (0 478-cm test section)						
dry	90	75	46	5 4	5 2	4 3
wet	57	38	16	11 4	10 0	13 3

^aRTD strength is the strength of specimens tested at room temperature without prior moisture conditioning.

TABLE 10—*Typical tensile, compressive, and in-plane shear strength values versus temperature*

Specimen Type	Percent of RTD Strength ^a		Coefficient of Variance	
	204°C/Wet	316°C/Wet	204°C/Wet	316°C/Wet
0° tension	97	92	6.9	6.4
90° tension	95	84	5.2	11.0
90° compression	48	17	18.7	26.2
± 45° in-plane shear (max shear)	83	46	5.3	12.9

^a RTD strength is the strength of specimens tested at room temperature without prior moisture conditioning

Absorbed moisture caused significant reduction in elevated temperature resin-dominated properties. This reduction in mechanical properties was significant even at moderate levels of moisture, such as 40 to 50% of saturation at the time of specimen failure. Elevated temperature exposure, even without prior moisture conditioning, also significantly reduced mechanical properties, especially at temperatures greater than 204°C.

Hot/Wet Testing

Hot/wet conditions can be obtained by graphite/PMR-15 laminates during service, therefore, the effects of these conditions must be known and considered in design. The effects of hot/wet conditions can be evaluated with coupon specimen tests using proper equipment and heating techniques. These conditions cannot be achieved with standard environmental test chambers, but can be achieved with high velocity hot air. To verify that the hot/wet test condition was achieved, specimen moisture content must be measured before and after specimen testing.

Hot/wet specimens should be heated quickly to retain sufficient moisture at the time of failure. This is more easily accomplished by heating only the specimen test section. One can determine the moisture content of a specimen test section after failure by heating a representative untabbed traveler with the specimen heat cycle. It is assumed that the traveler would be from the same panel, would receive the same environmental conditioning, and would be approximately the length of the heated test section. It is also recommended that industry standards be developed for moisture level requirements and moisture verification techniques.

Conclusions

When designing graphite/PMR-15 structures several variables must be determined and considered including the following: temperature, moisture absorption, and load time history relationships. Design allowables should then be developed using appropriate moisture levels and test temperatures that simulate the service environment. The test methods, heating procedures, and instrumentation used should permit such a test program. Specimen geometry can also greatly affect test results as shown by this program's two compression specimen configurations and their test results.

Acknowledgments

The author gratefully acknowledges the contributions and cooperation of C. R. Foreman and D. W. Perreault during the C3000/PMR-15 test program and the review of this

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A Scientific Approach to Composite Laminate Strength Prediction

REFERENCE: Hart-Smith, L. J., "A Scientific Approach to Composite Laminate Strength Prediction," *Composite Materials Testing and Design (Tenth Volume)*, ASTM STP 1120, Glenn C. Grimes, Ed., American Society for Testing and Materials, Philadelphia, 1992, pp. 142-169

ABSTRACT A new method of predicting the strength of cross-ply fibrous composite laminates is based on expressing the classical maximum-shear-stress failure criterion for ductile metals in terms of strains. Starting with such a formulation for classical isotropic materials, the derivation is extended to orthotropic materials having a longitudinal axis of symmetry to represent the fibers in a unidirectional composite lamina. The only modification needed to represent those same fibers with properties normalized to the lamina rather than fiber is a change in axial modulus. A mirror image is added to the strain-based "lamina" failure criterion for fiber-dominated failures to reflect the cutoffs caused by the presence of orthogonal fibers. It is found that the combined failure envelope is now almost identical with the well-known maximum-strain failure model in the tension-tension and compression-compression quadrants, but is truncated in the shear quadrants. The new predictions refer only to fiber-dominated failures of efficient, well-interpersed, laminate patterns. This restriction permits the laminate strengths to be defined by only half as many material properties as were used in the past. Premature matrix failures are excluded from consideration for conventional fiber/polymer composites. The new failure envelope is compared with published biaxial test data.

KEYWORDS: laminates, strength prediction, fiber composites, failure

Failure or yield of metals has traditionally been characterized in terms of applied stresses. It makes little difference for isotropic materials whether such expressions are formulated in terms of stress or strain. However, it makes a tremendous difference in predicting failure of orthotropic materials such as fibrous composites. Indeed, the almost universal preference for a stress-based reference has handicapped failure prediction for composites for a quarter of a century. In stress formulations, it has been incorrectly assumed that the longitudinal and transverse properties of the unidirectional lamina, which have served as the basis of laminated composite analysis, are independent quantities that need to be specified separately for use in an interactive type of failure criterion.

If the formulation had been in terms of strain, the dependence between longitudinal and transverse strengths would have been apparent. The characterization of failure or yield in terms of strain rather than stress permits a single universal criterion to cover *all* materials, isotropic and orthotropic. In the case of fibrous composites, the current longitudinal tests (tension and compression) are sufficient to characterize correctly the fiber-dominated failures, while the measured transverse strengths, which have nothing to do with the failure of the fibers, can serve as the basis of a *separate* failure criterion for the resin matrix.

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This paper explains how failure can be characterized in terms of strain rather than stress. The conversion for isotropic materials—metal alloys, for example—is derived first to show the tie-in to established methods. The formulation in terms of strain is then extended to orthotropic materials. That nonsymmetric failure criterion for a unidirectional fibrous composite lamina is next simplified by superposition with an equivalent strength cutoff for any orthogonal layer of the *same* composite material since, for most practical structural laminates, some orthogonal fibers will always be present.

The reduced envelope is then recognizable as a truncation in the in-plane shear quadrants of the well-known maximum-strain failure model. Whenever there are orthogonal fibers present, such an envelope is not an arbitrary truncation, but is directly equivalent to the generalized maximum-shear-stress failure criterion developed earlier by the author. But now it is simplified to the point where it should be more appealing to designers, and more importantly, it is so formulated that there is no need to specify any transverse properties. Consequently, researchers and analysts need not be skilled in the art to make realistic laminate strength predictions.

What Is Wrong with Composite Failure Models?

Despite much mathematical elegance, most published “failure theories” for composites are simply meaningless curves drawn through unrelated data points. They suffer from two basic errors. The first is an assumption that it is permissible to homogenize a mixture of fibers and resin *before* prescribing the failure criterion. Only a fiber, the matrix, or an interface can fail, and a separate criterion is needed for each case. The “lamina” cannot fail, it is merely a convenient mathematical concoction with which to calculate elastic constants (stiffnesses). The second common error is the malpractice of insisting on a single, smooth, continuous function to describe failure, even when it is known that the mode of failure changes with the state of stress.

Moreover, it is often erroneously assumed that the same characterization of a unidirectional lamina tested in isolation must apply equally when the same lamina is embedded in a cross-ply laminate. This would be true only if the same constituent failed in both situations. However, with many structural fibrous composites, this condition is not satisfied. The conventional characterization of $0^\circ/90^\circ$ laminates made from unidirectional tape is unjustifiably different from the characterization of the same fibers in woven fabric. A cross-ply tape laminate is assumed to be subject to the same matrix cracking as can be observed in a 90° tension test of an entirely 0° laminate, yet the phenomenon is ignored in woven fabrics containing the same fibers and resin because such failures, even if they did occur, do not result in catastrophic failure of the laminate. Failure does not occur until the much higher load needed to fail the fibers is achieved. It is not sufficient to “verify” a composite failure model by assessing multiple states of combined stress at the lamina level. The *predictions* of the model at the cross-ply laminate level must be verified instead.

Unreinforced concrete is an excellent vehicle for exposing some of these errors in other methods for predicting the strength of fibrous composite structures. With reference to Fig. 1, the failure of concrete under in-plane loads ($\sigma_z = 0$) can be characterized by a ductile shear mechanism everywhere except in the tension-tension quadrant and adjacent areas, where brittle fracture occurs long before sufficiently high stresses could be developed to cause a shear failure.

But, if one were to apply conventional composite failure theories such as those of Tsai-Hill [1], Tsai-Wu [2], or Hoffman [3] to the measured unidirectional tensile and compressive strengths, the failure envelope would look like the ellipse shown in Fig. 1 with obviously no

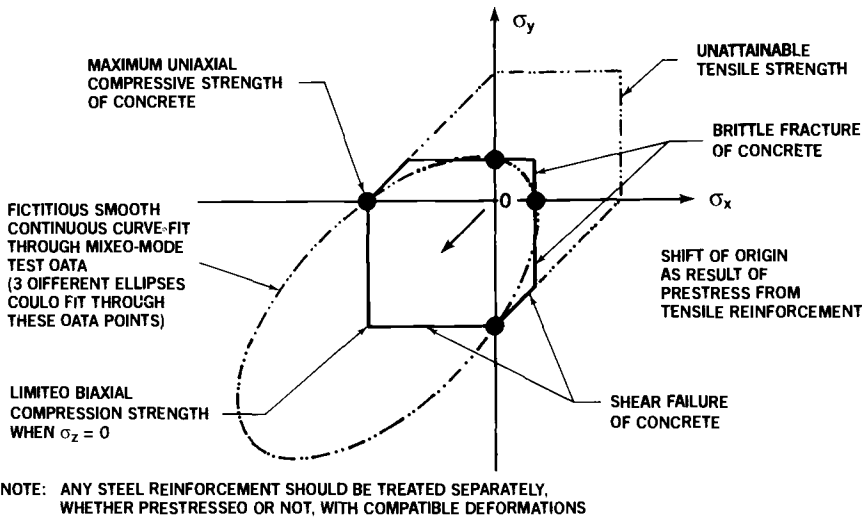
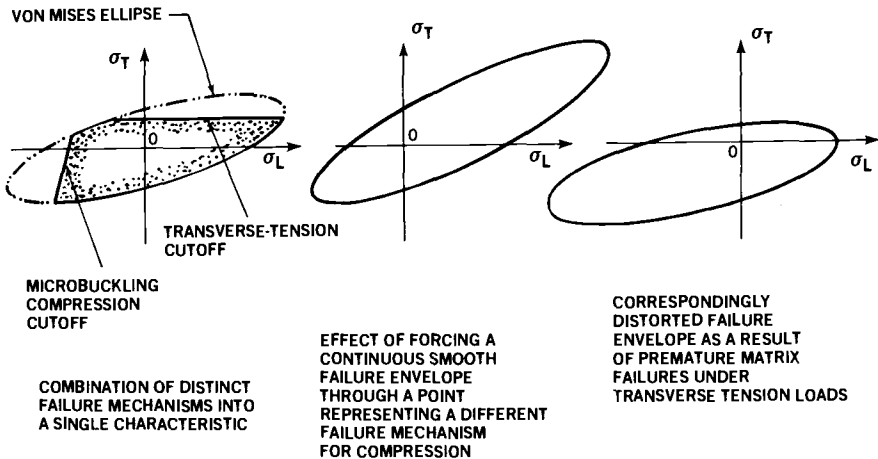


FIG. 1—Unreinforced concrete as an example of inadequacies of tensor/polynomial failure “theories” in the presence of mixed-mode failures.

relationship to reality. Yet, this is precisely the way such theories would be applied to a 0°/90° layer of woven fabric.

Worse, when such theories are used to construct failure envelopes through the measured longitudinal and transverse tensile and compressive strengths of unidirectional fibrous composite laminae, one is faced with the problems shown in Fig. 2. Even a minor change in measured strength at one location causes incorrect predicted changes throughout *all* of the failure envelope except for the other three data points. Logic should require kinked failure envelopes for multiple failure modes, as shown at the left of Fig. 2.

In his original work on plasticity, Hill [4] devised a characterization for equivalent combinations of triaxial stresses which would all result in the yielding of slightly orthotropic mate-



rials such as metal extrusions and rolled plates. His failure envelopes can be smooth because he confined his attention to a *common* failure mechanism and a *homogeneous* material. Unfortunately, many who attempted to adapt his techniques to fibrous composites have failed to recognize these implied restrictions.

Adaptation of the classical ductile shear (Tresca) yield criterion for isotropic metals to fiber-dominated failures of orthotropic materials such as carbon fibers will now be explained. It is acknowledged that any matrix-dominated failures would need additional failure criteria before they could be considered. Fortunately, this restriction is not severe since skillful designers avoid all bad laminates in which a poor fiber pattern prevents efficient use of the fibers. Nevertheless, particularly for compression-dominated loads, it is appropriate to provide a mechanism for *locally truncating* the basic failure envelope *without* making any changes in stress domains in which any alternate failure mechanism should *not* have any effect. Characterizing the generalized maximum-shear-stress failure criterion in the strain plane, rather than by the more conventional stress-based reference, not only provides a physically plausible model for the basic failure envelope, but it also permits the simple superposition of other plausible failure modes whenever they are appropriate.

The Maximum-Shear-Stress Failure Criterion in Terms of Strain for Isotropic Homogeneous Materials

The classical maximum-shear-stress (Tresca) yield or failure criterion for homogeneous isotropic materials is shown in the form of Mohr stress and strain circles in Figs. 3 and 4, respectively. Any combination of stresses having principal stresses predicted to differ by more than twice the critical shear stress shown in Fig. 3 would represent failure. The corresponding criterion on the strain plane in Fig. 4 is the critical shear strain γ_{crit} . The shear stresses τ and strains γ are related by the particularly simple relation

$$\tau = G\gamma \quad (1)$$

because of the isotropy of the material, for which G is the shear modulus.

This same criterion can alternatively be expressed on a different stress plane in the form shown at the left of Fig. 5, in which σ_x and σ_y are in-plane direct stresses. For the special case of in-plane stresses of interest for composite laminates, with no normal stress ($\sigma_z = 0$), the

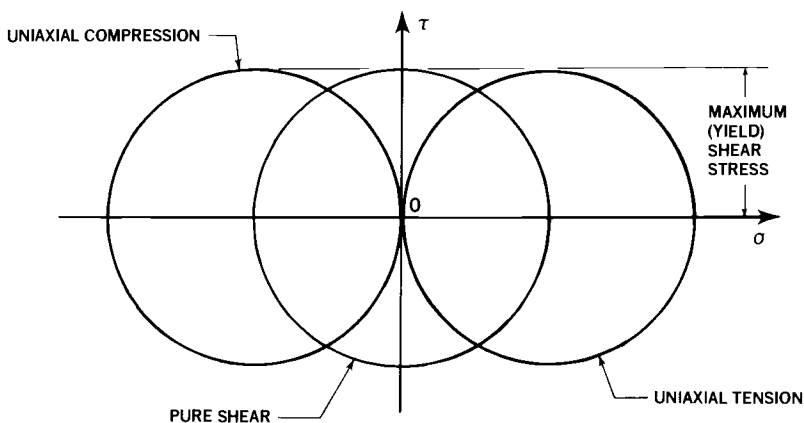


FIG. 3. Mohr stress circles for isotropic materials.

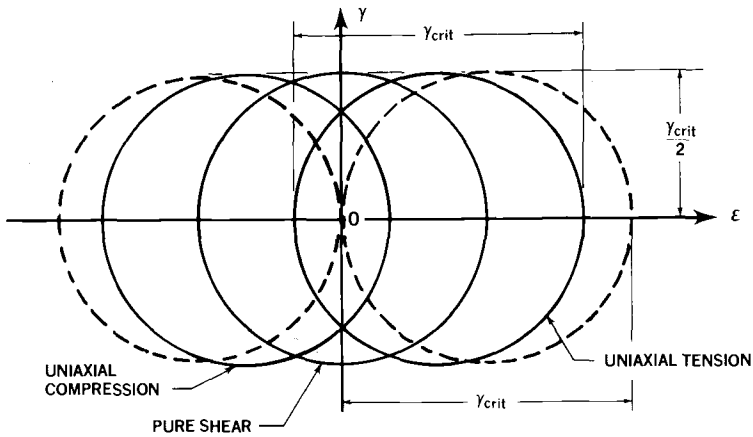
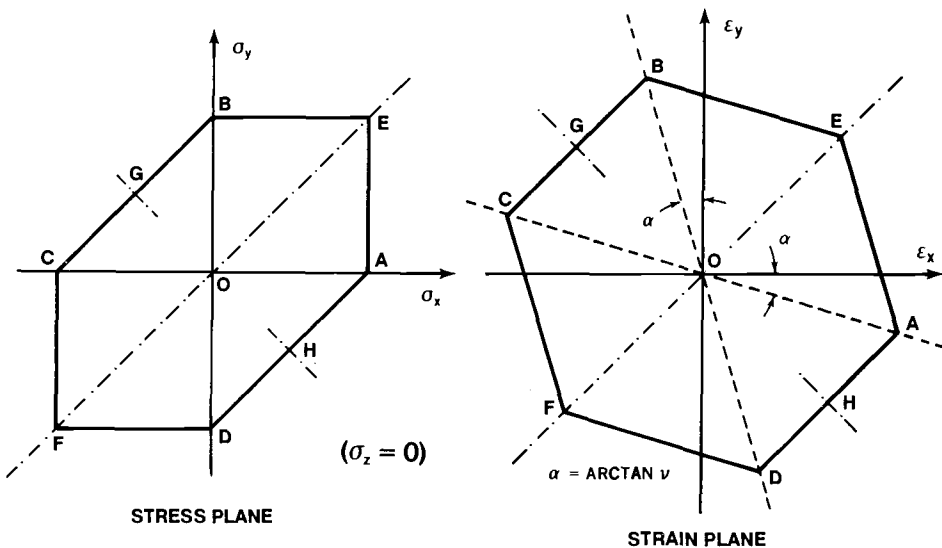


FIG. 4—Mohr strain circles for isotropic materials.

equivalent criterion in the $\epsilon_x - \epsilon_y$ strain plane is shown on the right of Fig. 5. Note that, although the strains are referenced in the x - y plane, the critical conditions can occur in any of the three planes (x - y , x - z , and y - z) depending on the state of stress.

For instance, at the uniaxial tension Point A, critical shear stresses are developed in the x - y and x - z planes, while the same critical shear stresses are developed in the x - z and y - z planes at the equal biaxial tension Point E. Similarly, for uniaxial compression at Point D, the critical planes are x - y and y - z ; at the biaxial compression Point F, the critical planes are x - z and y - z as at Point E. Points located *off* the corners are critical in *one* plane only. For instance, the equal and opposite biaxial tension and compression Points G and H are critical in the x - y plane only. The reason why the in-plane direct stresses for uniaxial and equal biaxial stresses are the same is that this is the inevitable consequence of failure (or yield) being governed by a common critical shear stress (or equivalent *difference* between principal stresses). Both failure



envelopes in Fig. 5 are *entirely* defined completely by this *single* starting point. The uniaxial and biaxial points should not be looked upon as being individually determined by *separate* assumptions or criteria. (This point needs to be clarified here because, at the *laminate* level, cross-plyed fibrous composites can have quite different uniaxial and biaxial strengths depending on the particular lay-up pattern. Nevertheless, at the individual unidirectional lamina, or more strictly fiber level, the critical shear stresses are identical at both the uniaxial and equal biaxial stress points.)

The letters at each corner of the hexagons in Fig. 5 correspond precisely. The various states of stress are identified in Fig. 6 on the strain plane. For example, Point A represents uniaxial tension in the x direction:

$$\begin{aligned}\sigma_y = \sigma_z = 0, \sigma_x = \sigma_0 \\ \epsilon_x = \epsilon_0 = \sigma_0/E, \epsilon_y = \epsilon_z = -\nu\sigma_0/E\end{aligned}$$

where ν is the Poisson's ratio for this isotropic material and E the Young's modulus.

The uniaxial strain at failure is given by

$$\epsilon_0 = \gamma_{\text{crit}}/(1 + \nu) \quad (2)$$

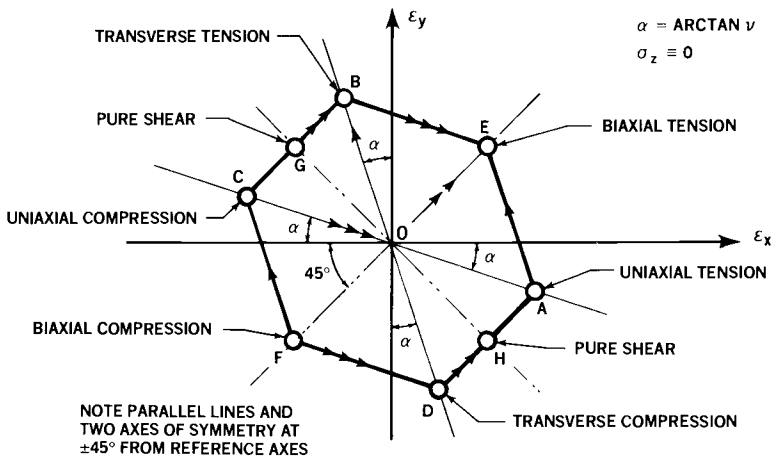
and it is now apparent that the critical shear strain, or principal direct strain *difference*, is the primary quantity and that the axial strain ϵ_0 is secondary.

Point F represents equal biaxial compression; $\sigma_z = 0, \sigma_x = \sigma_y$. The critical strain at the equal biaxial stress points is defined by

$$\epsilon_x = \epsilon_y = \left(\frac{1 - \nu}{1 + \nu} \right) \gamma_{\text{crit}} = (1 - \nu)\epsilon_0 \quad (3)$$

Points G and H denote equal and opposite in-plane stresses and strains; $\sigma_z = 0, \sigma_x = -\sigma_y = \pm\sigma_0/2$. The corresponding in-plane strains at these opposite biaxial, or shear, stress points can be seen to be as follows

$$\epsilon_x = -\epsilon_y = \pm\gamma_{\text{crit}}/2 = \pm(1 + \nu)\epsilon_0/2 \quad (4)$$



Before generating equivalent failure criteria for orthotropic materials, it is useful to interpret some of the less obvious lines in Fig 6 The radial Lines 0A, 0B, 0C, and 0D are easily interpreted as uniaxial load lines Along the Line AE, the stress σ_y is increased in such a way that the stress *difference* $\sigma_x - \sigma_z$ is held constant because the application of σ_y stresses alone cannot possibly induce any stresses in the x - z plane—only compatible Poisson strains Therefore, the axial and normal strains ϵ_x and ϵ_z must decrease by an amount $\nu\epsilon_y$ as σ_y increases Since σ_z is zero at Point A, σ_x is constant along the Lines AE and CF in both Figs 5 and 6 Similarly, σ_y is constant along the Lines CE and BF

Again, it is appropriate to observe that Figs 5 and 6 characterize the states of stress or strain in the x - y plane at failure, even when the critical conditions are developed in the x - z or y - z planes

The lengths of all the sides of the hexagon in Figs 5 and 6 would be identical only for one particular value of the Poisson's ratios That value is obtained by equating the lengths of the usually different sides, as in the following equation

$$\frac{\sqrt{(1 + \nu^2)}}{(1 + \nu)} = \sqrt{2} \left(\frac{1 - \nu}{1 + \nu} \right)$$

$$(1 + \nu^2) = 2(1 - 2\nu + \nu^2)$$

or

$$1 - 4\nu + \nu^2 = 0 \tag{5}$$

The result, $\nu = 0.2679$ is typical for metals, so that the sides of most hexagons would be nearly equal

The (Isotropic) Maximum-Shear-Stress Failure Criterion Formulated in Terms of Strains for Orthotropic Laminae

The next step is to generalize the preceding analysis to orthotropic layers The failure criterion to be derived should actually refer to the fiber, for fiber-dominated failures, *not* the lamina However, the fibers and matrix obviously share a common strain along the fiber axis For the usual case, in which strong, stiff fibers are embedded in a soft matrix, it is possible to bypass the micromechanical calculations relating the transverse strains of the fiber to those of the lamina because of the greater criticality of the fiber failure criterion Such a simplification may not be realistic for some of the more exotic advanced composites, and a more comprehensive theory would be needed then

It is assumed again that there are no normal or through-the-thickness shear stresses acting on the laminate Using the Subscripts L , T , and N to denote the longitudinal, transverse (in-plane), and normal directions, respectively, the strain-stress relations for transversely isotropic materials read as follows when the axis of symmetry is longitudinal

$$\epsilon_L = \frac{1}{E_L} (\sigma_L - \nu_{LT}\sigma_T) \tag{6}$$

$$\epsilon_T = \frac{1}{E_T} (\sigma_T - \nu_{TL}\sigma_L) = -\nu_{LT}\sigma_L/E_L + \sigma_T/E_T \tag{7}$$

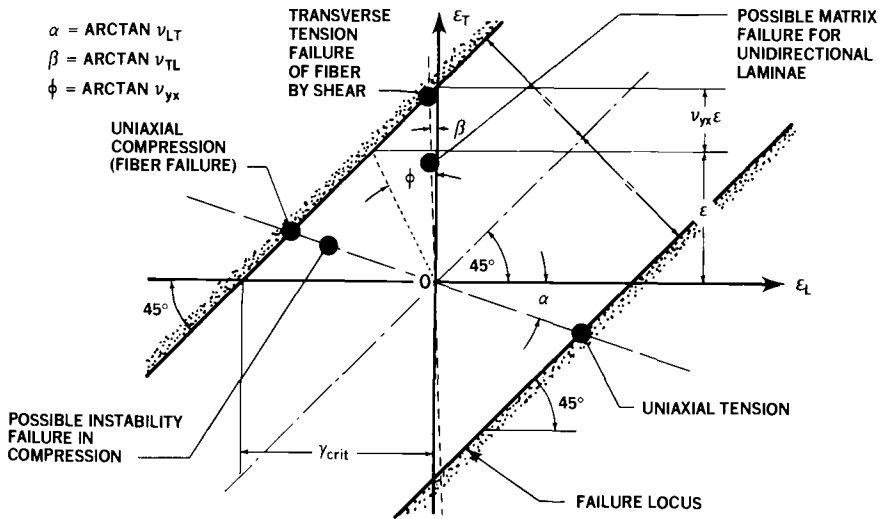


FIG. 7—Problems with experimental confirmation of lamina failure model.

and

$$\epsilon_N = \frac{1}{E_T} (-\nu_{TL}\sigma_L - \nu_{TN}\sigma_T) \quad (8)$$

For a uniaxial load along the fibers, with neither lateral nor normal applied stress, the axial strain in both the fiber and lamina will be ϵ_0 at failure, and the associated lateral and normal strains will be $-\nu_{LT}\epsilon_0$ in the lamina. The corresponding orthogonal strain in the fiber, which can be presumed to have a different Poisson's ratio from the homogenized lamina, is undefined. However, for typical carbon-epoxy composites, the similarity of the Poisson's ratio ν_{LT} for the unidirectional lamina and an isotropic resin matrix implies that the corresponding Poisson's ratio for the fiber must also be similar. If a substantial difference is expected between the Poisson's ratios for the fiber and monolayer, one would need to complete the micromechanical analyses relating the two before proceeding with the formulation of the failure criteria. And the appropriate distinction would need to be made for all states of stress being considered.

Figure 7 shows the traces of these uniaxial load lines, in tension and compression, on the $\epsilon_L - \epsilon_T$ in-plane strain plane. If the tensile and compressive strengths are the same, it is possible to locate the 45° sloping shear-failure lines on which the uniaxial failures are represented by individual points. If the tensile and compressive strengths differ, the numerically greater value defines both shear-failure lines, as shown in Fig. 7, while the lesser strength defines a cutoff caused by some other failure mechanism, such as microbuckling under compression.

For a uniaxial in-plane load perpendicular to the fibers, a different Poisson's ratio, ν_{TL} , is involved, and the uniaxial load line is no longer symmetric as it was for isotropic materials in Fig. 6. The traces of this unidirectional load condition can be added to Fig. 7, and wherever they cross the shear-failure lines denotes the points of failure of the fibers.²

²With unidirectional laminae, the matrix may fail prematurely under transverse tension at a lower strain, but that is normally suppressed for any typical cross-ply structural laminate and may be considered as a special case.

It is apparent that the same critical shear strain holds throughout the entire region of interest in Fig. 7 since the sum of longitudinal and orthogonal strains is always

$$\epsilon + \nu\epsilon = \gamma_{crit} \tag{9}$$

no matter what the value of the Poisson's ratio ν for any particular cross-plyed laminate.

To complete Fig. 7 and make it equivalent to Fig. 6, constant-stress lines are added through the uniaxial stress points associated with shear failures of the fibers. Or, if there are premature failures under other modes, the additional lines must be placed where the missing shear failures would have occurred. Consider, for example (Fig. 7), the addition of transverse stress to the point of uniaxial tension loading while holding the longitudinal stress in the fibers and lamina constant without altering the stress-free state in the normal direction. Doing so would require that the additional strains be in the same relation as those following the purely transverse tension line in Fig. 7. Along that line, with a slope defined by ν_{TL} , only the transverse stress varies. There are no incremental longitudinal and normal stresses, so no shear stress can be induced in the L - N plane. Likewise, the missing line from the transverse tension to the biaxial tension points must be parallel to the original longitudinal tension line in Fig. 7. The entire envelope is shown completed in Fig. 8 for the special case in which the tensile and compressive strengths of the unidirectional laminae are equal. Unlike Figs. 5 and 6, which are doubly symmetric, Fig. 8 is skewed because of the difference between the two in-plane Poisson's ratios. The third Poisson's ratio is involved in establishing the compatible normal strains, but does not appear in the in-plane failure envelope shown in Fig. 8.

Figures 9 and 10 show the effect of fiber-dominated cutoffs at either end of the failure envelope. The cutoff representing microbuckling of the new (small-diameter) high-strain carbon fibers is drawn perpendicular to the strain axes because compressive buckling is not usually sensitive to orthogonal stresses. The other cutoff, representing a lower tensile than compressive strength, is drawn parallel to the unidirectional load line since it is more likely to be a stress-imposed limit than not.

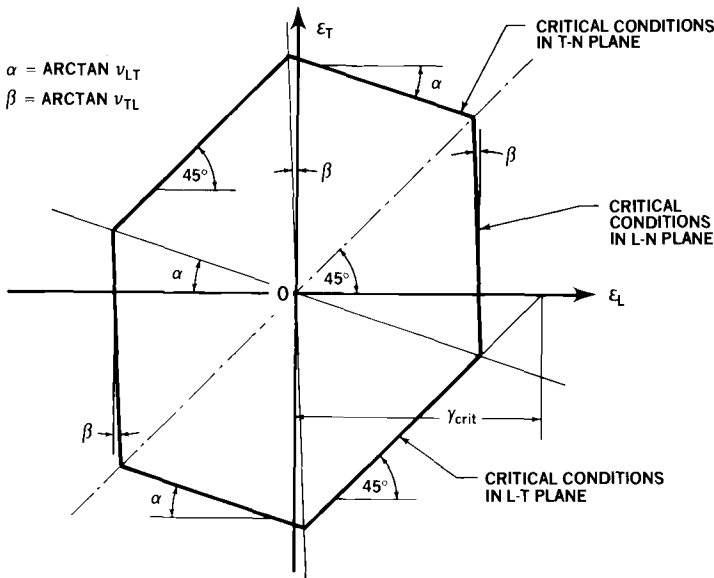


FIG. 8. Shear failure envelope for orthotropic materials having equal tensile and compressive strengths. Copyright © 1994 by ASME. Downloaded/printed by North Carolina A & T State (North Carolina A & T State) pursuant to License Agreement. No further reproductions authorized.

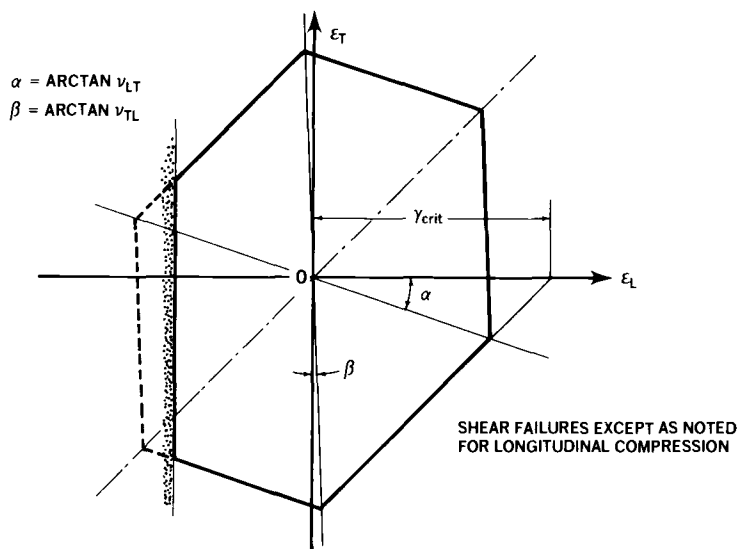


FIG. 9—Superposition of compressive instability limit on shear-failure envelope.

Figure 11 shows a postulated cutoff for matrix-dominated transverse tension failures. Since this would appear to be a stress rather than strain limit, the cutoff is drawn parallel to the constant transverse stress line on the basis that any longitudinal load would not induce any stress in the T - N plane.

Figures 8 through 10 refer to a fiber-dominated unidirectional lamina and could be used directly as a ply-by-ply failure criterion for analyzing cross-plyed structural laminates. However, many simplifications ensue from reinterpreting this criterion in the context of laminates *before* it is used as a strength check, as discussed in the next section.

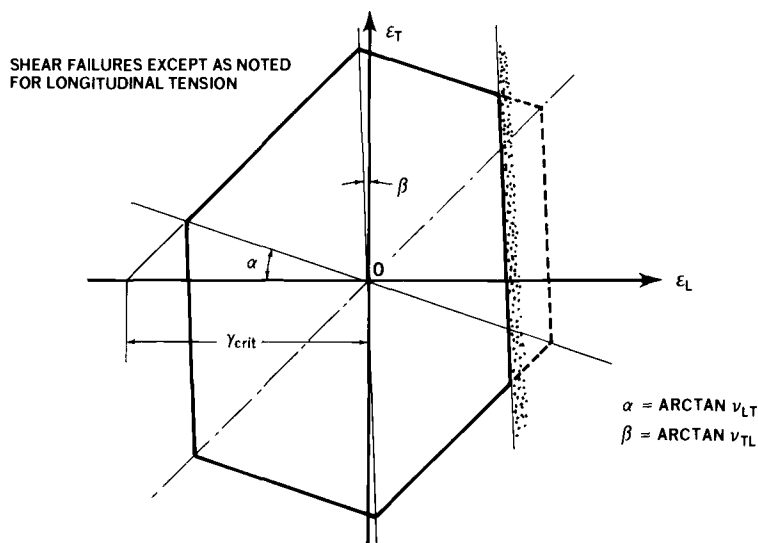


FIG. 10—Superposition of brittle fracture cutoff on shear-failure envelope.

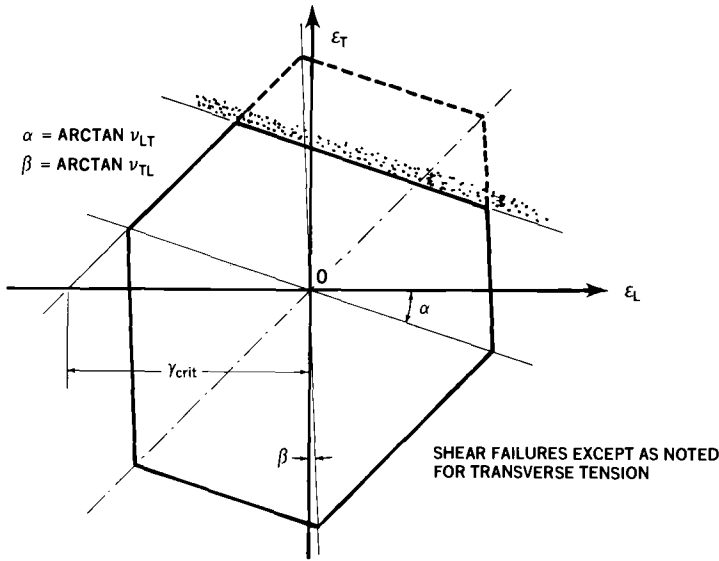


FIG. 11—Superposition of transverse-tension cutoff on shear-failure envelope.

The “Maximum-Shear-Strain” Failure Criterion Formulated for Cross-Plied Structural Laminates

It is now apparent that the generalization of the classical maximum-shear-stress failure criterion for ductile metals is actually most conveniently expressed as a maximum-shear-strain criterion for orthotropic materials, encompassing the classical hypothesis as a special case. Care is needed to exclude shear strains caused by factors other than stress, as, for example, nonuniform thermal expansion or swelling as a result of absorbing moisture. Even when attention is paid only to the mechanically caused shear strains, there are still some difficulties since some of the Poisson-induced strains have no corresponding stresses, as discussed in the next section.

In design of composite aircraft structures, it is customary to have a minimum percentage of fibers in *all* of the four standard directions: 0, +45, 90, and −45°. That being the case, the greater strains in Figs. 8 through 10 in the transverse rather than longitudinal direction would be truncated by longitudinal failures in orthogonal plies. So, if there were really orthogonal plies for each fiber direction, these figures could be simplified with no loss of accuracy. This has been done in Fig. 12 by adding a mirror image to Fig. 8 about the biaxial strain diagonal axis and taking the smaller strength cutoffs for each segment of the envelope. The new figure is doubly symmetric, just as Figs. 5 and 6 were for isotropic materials. More importantly, the corner points are defined by the Poisson’s ratios for the unidirectional laminae on which the failure envelope is based. This is of vital importance because, as shown below, those corner points remain fixed for *all* cross-ply laminates, regardless of the Poisson’s ratios of the actual laminate.

Because Fig. 12 is drawn in strain space rather than stress space, it remains doubly symmetric even if the percentages of 0 and 90° fibers are *not* the same. The two laminate Poisson’s ratios ν_{xy} and ν_{yx} would vary with the proportion of 0 and 90° fibers, but the failure envelope would *not* do so.

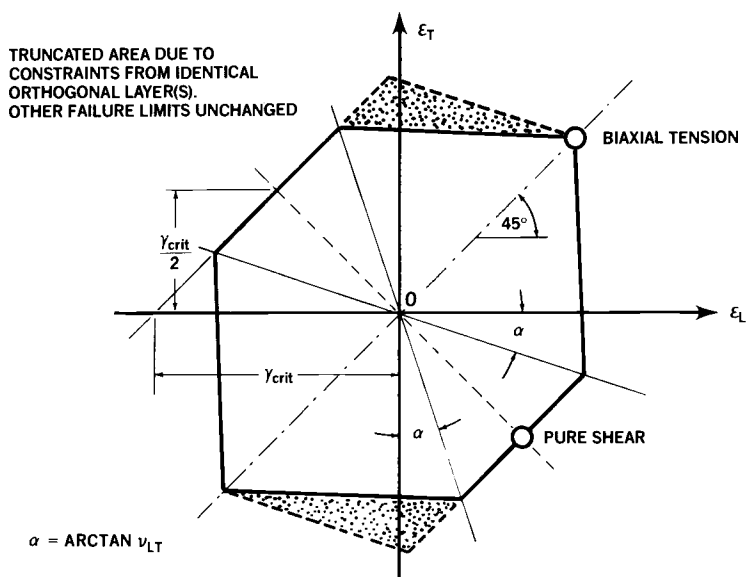


FIG. 12—Superimposed failure envelopes for orthogonal layers of orthotropic materials.

Consider the case of a laminate containing equal percentages of 0 and 90° fibers, with no $\pm 45^\circ$ fibers, for which both in-plane Poisson's ratios are approximately 0.05. Figure 13 shows where the uniaxial load lines in the longitudinal and transverse directions would intersect the failure envelope in Fig. 12. No intersections fall on the 45° sloping shear-failure lines, and the associated maximum shear strain $(1 + \nu_{xy})\epsilon_0$ appears insufficient to have caused failure.

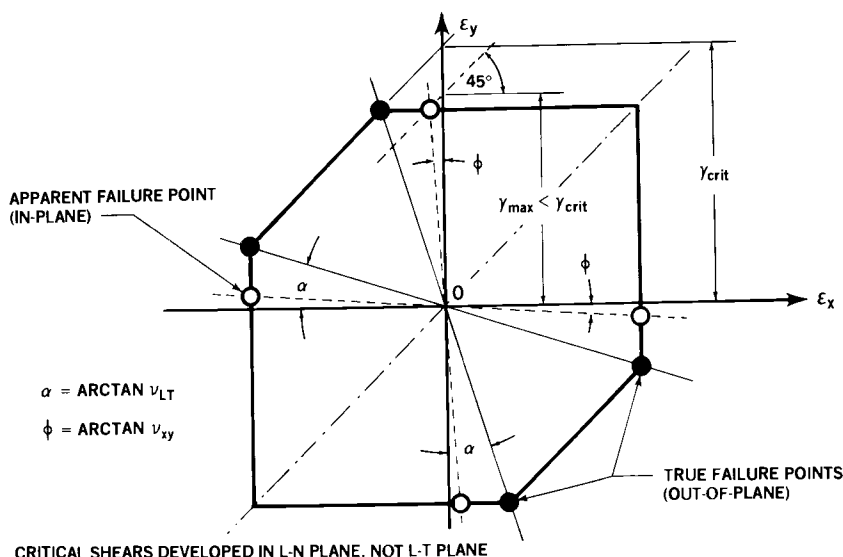


FIG. 13—Uniaxial failure points for $0^\circ/90^\circ$ cross-plyed laminates with low Poisson's ratios.

The explanation of this apparent anomaly is that failure is actually caused by the stress states in the planes perpendicular to the fibers involving the normal rather than in-plane direction. A longitudinally stretched all 0° laminate contracts as much laterally as it does through the thickness. However, the presence of the orthogonal in-plane fibers in the $0^\circ/90^\circ$ laminate restricts the in-plane lateral contraction, and hence, the in-plane shear stress developed in both the lamina and fiber, so that this plane becomes less critical than the one defined by the fiber axis and the normal direction. Failure is still predicted to occur on the diagonal shear failure line in Fig 7, only now the failure point is located on Fig 13 by a radial line with a slope defined by ν_{LT} rather than by ν_{xy} . This is generally true for all laminate patterns for which $\nu_{xy} < \nu_{LT}$ for the unidirectional lamina on which Fig 7 is based.

What is happening may better be understood by drawing the appropriate Mohr circles. Yet, for all practical purposes, the lines with slopes defined by ν_{xy} in Fig 13 and ν_{yx} (not shown) also correctly identified the failure strains ϵ_0 for the $0^\circ/90^\circ$ laminate. Similarly, the biaxial failure strains, which obviously are not associated with in-plane shear strains, are also correctly located.

If, for this same $0^\circ/90^\circ$ laminate, biaxial rather than uniaxial loads were to be applied, one could easily envisage a little transverse compression combined with a predominantly longitudinal tensile load so that the transverse strains were made to coincide with those developed by a uniaxial load on a unidirectional lamina. In such a case, failure would occur precisely on the corner point in Fig 8 because those would be the applied strains in the laminate, only they would have been developed partly by orthogonal loads instead of entirely by Poisson contractions.

Likewise, one could apply equal and opposite loads in the 0 and 90° fibers. This would induce equal and opposite strains, and failure would be predicted to occur at an axial strain of $\gamma_{crit}/2$, as noted in Fig 12. Indeed, that is the point at which *all* purely in-plane-shear failures will occur regardless of the particular Poisson's ratios for any particular cross-plyed laminate.

Consider next a nonstructural cross-plyed laminate consisting of 33% of the fibers in the 0° direction and the remaining 67% shared equally between the $\pm 45^\circ$ directions. For this laminate, the Poisson's ratio ν_{xy} will be about 0.67, much higher than the typical 0.25 for a unidirectional lamina and the 0.33 for a quasi-isotropic laminate. In the absence of 90° fibers, one should ideally draw the unidirectional load lines on Fig 8 rather than Fig 12. The other Poisson's ratio ν_{yx} should be about 0.2. Figure 14 shows these radial uniaxial load lines added to Fig 8 rather than Fig 12 because of the absence of 90° plies. It is significant that, in the longitudinal direction, the shear-failure line is intercepted at a *lower* axial strain than was the case for the unidirectional lamina. The critical difference between strains does occur *in-plane* in this case, and the longitudinal strain at failure is expected to be less than that of the unidirectional lamina in the following ratio

$$\epsilon_{\text{laminate}} = \epsilon_{\text{lamina}} \times (1 + \nu_{LT}) / (1 + \nu_{xy}) \quad (10)$$

In this case, the reduction is to only 75% of the nominal value. Thus, this laminate without any 90° fibers is not only impractical because of its disproportionately small transverse (90°) strength and stiffness, but also because it does not allow the load-carrying 0° fibers to work to their full extent.

Actually, the loss of efficiency is probably overestimated slightly because no distinction has been made between the lateral strains in the fiber and the lamina. With a truly soft matrix, the fibers would not compress quite as much as the lamina would shrink, with a lower lateral strain experienced by the fibers, the appropriate values of transverse strain would need to be reduced, and doing so would violate the standard assumption that plane sections remain plane throughout the laminate. The maximum-strain failure model for composites may be looked upon as

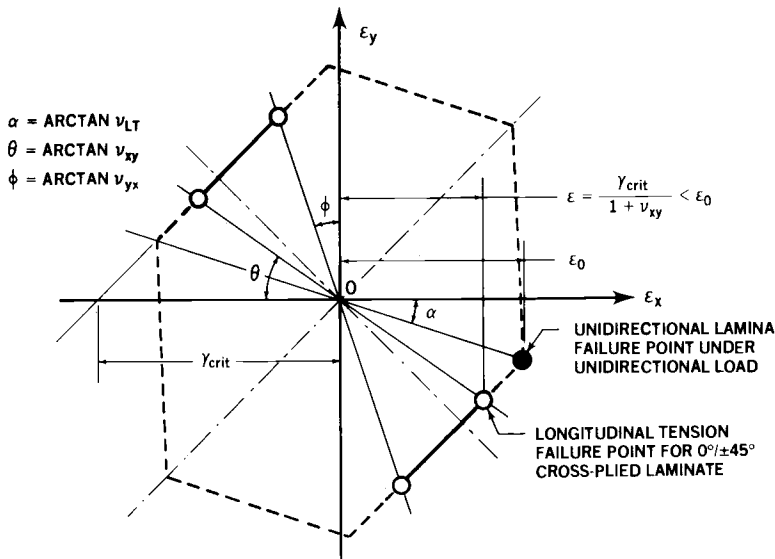


FIG. 14—Uniaxial failure points for $0^\circ/\pm 45^\circ$ laminates with high Poisson's ratios.

an overcorrection for this problem inasmuch as no allowance is made for any loss of fiber strength caused by a lateral stress of the opposite sign to the axial stress in the fiber. This issue can be resolved only by micromechanics; the same problem also exists with all the pseudo-scientific laminate strength theories.

For a quasi-isotropic laminate, the strain to failure would be predicted by Eq 10 as 94% of that of the unidirectional lamina. Again, in reality it would be slightly higher, but with good test specimens and test technique, it should be possible to detect a statistically significant small reduction from the unidirectional value.

If one were to test a cross-plyed laminate that had the *same* Poisson's ratios as the unidirectional lamina because of the particular mixture of 90° and $\pm 45^\circ$ fibers, the longitudinal strains to failure would logically be identical, and the unidirectional load lines would be the same as in Fig. 8.

Now, even though Fig. 12 can be applied to woven fabric laminates as well as to cross-plyed tape ones, the Poisson's ratios defining the corner points *must* be those of the *unidirectional* lamina, which may be difficult to measure on a woven fabric. However, they can be easily related and the required values can be deduced reliably from measurements of the Poisson's ratio of a $\pm 45^\circ$ woven laminate. (The direct use of the $0^\circ/90^\circ$ laminate is not recommended for this purpose since its Poisson's ratio is so small as to be difficult to measure.)

The failure envelopes in Figs. 8 through 10 would be applied, in turn, to the $0^\circ/90^\circ$ combination of fibers and the $\pm 45^\circ$ set. This would be necessary and sufficient to assess the strength of such laminates for fiber-dominated failures of in-plane loads. These are not unreasonable limitations since resin matrices are traditionally so weak that designers should always avoid applying direct transverse shear loads. Also, normal pressure loads must always be small in comparison with the in-plane stresses for any thin-shelled structures, although this might not be the case for deep submersible vehicles. Further, if the matrix is so weak as to prevent the fibers from developing their full strengths before failure of the laminate under in-plane loads, the matrix or the fiber pattern should be changed rather than the analysis method.

The Truncated Maximum-Strain Failure Model

The transverse monolayer Poisson’s ratio ν_{TL} is only about 0.025 for the usual case of carbon, boron, or glass fibers embedded in polymeric matrices like epoxies, polyester, and phenolics. It is not unreasonable under such circumstances to approximate the almost horizontal and vertical lines joining the uniaxial and biaxial load points in Fig. 12 by exactly horizontal and vertical lines, as in Fig. 15. If that is done, the new failure criterion appears as a close approximation of the now classical maximum-strain failure model for fibrous composites (see Ref 5), but with the latter’s overestimated strengths in the shear quadrants eliminated. A sample of these unconservative predictions is presented in Fig. 16.

Obviously, this type of simplification should not be expected to be valid for such composites as whisker-reinforced metal matrices for which the assumption of strong and stiff fibers and a soft and weak matrix would no longer be valid. But many of the structural applications of fiber/polymer composites would be well characterized by the simple failure models shown in Fig. 15. Basing the application of the new failure criterion—that of the generalized maximum-shear-stress recommended by the author in Refs 6 through 8—on an improvement in the well-known maximum-strain model for analyzing composite structures should have the effect of making the physics of the phenomena easier to understand. And, since it is now seen to be directly equivalent to the truly classical ductile-yield failure theory for metals, few designers and analysts should feel the kinds of insecurity and concern that have been associated with earlier composite failure criteria.

Nevertheless, the author again cautions that, until a proper two-phase failure theory is developed to characterize the in-situ strengths of the fibers and matrix separately, even this theory is capable of misapplication. It is unlikely to be improved upon for general purpose analyses of today’s fiber-reinforced polymer matrices, except for Kevlar (aramid) fibers that exhibit more than the two modes of failures observed in carbon fibers, so long as designs exclude the excessive degrees of orthotropy and interlaminar weaknesses associated with clustering too many parallel fibers together (either in layers or bundles). The problem is that this new theory, like all prior theories, is capable of being misapplied without any restrictions on permissible fiber patterns and without warning the user that he should not believe the predictions. This problem is critical for laminate selection by computer optimization codes, which are not programmed to reduce the laminate strengths for matrix-dominated failures, merely because no one has derived all the necessary equations.

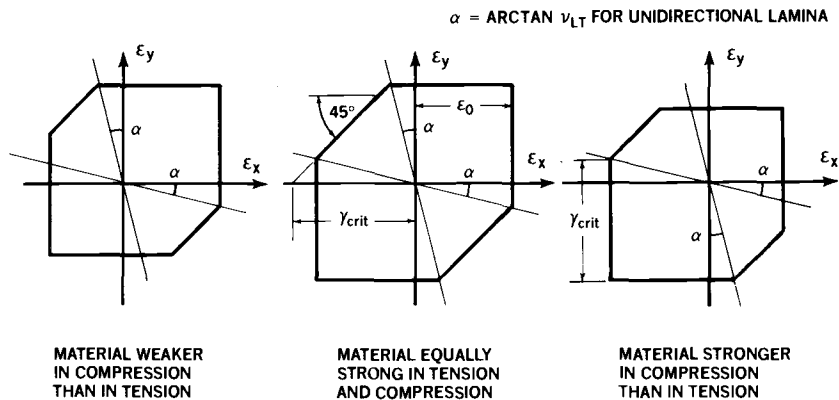


FIG. 15—Truncated maximum-shear-strain failure models for fiber-dominated failures of cross-plyed

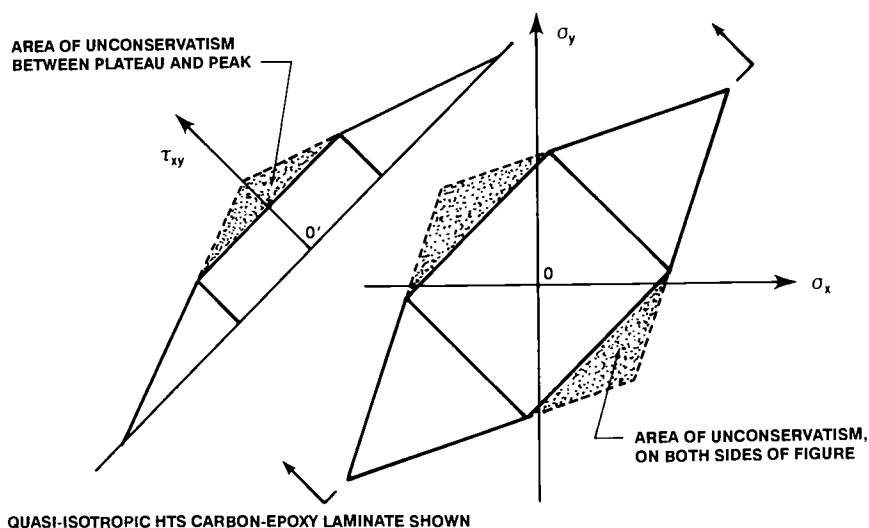


FIG. 16—Unconservatism of untruncated maximum-strain failure model.

Ironically, a better theory for state-of-the-art fiber/polymer composites should merely discourage the use of impractical fiber patterns that have already been identified on the basis of accumulated experience but are still very much in vogue because oversimplified analyses have failed to condemn them. A better theory should not be expected to improve the strengths predicted for practical fiber patterns. However, the author believes that such improvements, from the derivation of a truly two-phase theory with possibly interface failures as well, will be vital to the use of the more exotic composites on which research has already begun.

In comparison with the "generalized maximum-shear-stress" failure criterion coded in the BLACKART computer program (see Ref 8), the new strain-based formulation will predict the *same* composite laminate strengths whenever input data for BLACKART have been appropriately selected to make them represent the behavior of the unidirectional lamina within a cross-ply laminate, rather than in the isolated circumstances in which it has customarily been tested.

The real beauty of the present strain-based formulation is that there is no opportunity to enter transverse properties that are incompatible with the longitudinal properties. The only data needed are the lamina effective strain-to-failure in tension and compression and the Poisson's ratios of the lamina and laminate. Conversion to the corresponding strengths requires the addition of only the laminate stiffnesses, which have been calculated reliably for a long time.

Nevertheless, some residual aspects of this problem remain elsewhere in the analysis process. The elastic constants used to construct finite elements usually need to be based on effective (or secant) moduli rather than on initial tangent values. Otherwise, proper sharing of the load will not be achieved by the ply-by-ply decomposition of the internal stresses before the failure criterion is applied. An excessive transverse stiffness would cause any transverse plies to appear to be accepting excessive loads and therefore be even more likely to fail than would really be the case. On the other hand, the transverse stiffness of conventional fiber/polymer composites is so much less than the longitudinal stiffness that little error is likely to be introduced in the estimate of the load carried by the longitudinal plies. Thus, by eliminating the customarily premature transverse tension cutoff from the failure criterion, one will often

obtain correct estimates of laminate strengths in spite of the improper treatment. As shown in Ref 8, when transverse-tension failures were suppressed, doubling and halving the transverse-tension modulus had little effect on the failure envelope. Only when an artificially low transverse strain-to-failure was included in the analysis did the predictions become nonsensical.

It may appear that the author has made precisely the same mistake that he criticized others for, that is, homogenizing the fibers and matrix into a single orthotropic composite material. It is true that the material properties are expressed at the lamina level, but the new failure criterion is really a normalized expression of the failure of the fibers alone. Actually, one should perform micromechanical analyses of the lamina and its constituents to relate the Poisson contractions of the fibers to those of the lamina because the transverse strains of the fibers need not be the same as for the lamina. And it is possible that when this is done, pure-shear failures will be found to lie slightly off the -45° line in Fig 8, but still on the *same* shear failure line as for fiber failure under longitudinal tension or compression.

However, in the context of Fig. 15, if the longitudinal Poisson's ratios of the fiber and lamina differed significantly, an *equivalent* failure envelope could be prepared for the Poisson's ratio of the fiber. There would be a compensatory shift in the location of the uniaxial failure points, but the *form* of the failure envelope would not change at all because the transverse strains were expressed at the lamina rather than fiber level. (Obviously, the longitudinal strains must be the same except for edge effects not normally addressed by laminate analysis.) Nevertheless, this simplification is appropriate *only* for those laminate patterns that are associated with fiber-dominated failures. The precise two-phase analysis would be required for matrix-dominated failures.

As is explained later, to superimpose fiber and matrix failure criteria, it is first necessary to use a common strain base—that of the laminate and hence of each lamina—rather than the strains of the constituents. So, the failure envelopes should always be expressed at the lamina rather than constituent level. The net effect of a micromechanics analysis distinguishing between constituent and macrostrains would then appear as a change in slope of the line relating longitudinal to transverse strains within each lamina. The Poisson's ratio slopes given by ν_{LT} and ν_{TL} in Fig 8 would not change, but those defined by ν_{xy} and ν_{yx} in Figs 12 and 14 could. Likewise, the pure-shear point at the laminate level might not be precisely on the -45° sloping line when assessed from the point of view of either the fiber or matrix.

Effects of In-Plane Shear on the “Lamina” Failure Criterion

If the principal directions of stress or strain do not coincide with the fiber axes, a question arises as to the validity of the preceding failure criteria that makes no provision for such a situation. In the case of ductile *homogeneous* materials, the maximum-shear-stress failure (or yield) criterion in Fig 5 can *always* be used by a simple rotation of reference axes to coincide with the principal stress and strain directions, so this criterion is inherently complete. However, for fibrous composites, such a rotation of reference axes would invalidate the use of the simple stress-strain relations in Eqs 6 through 8. Many more terms would then be needed to characterize the stress-strain relations. The seemingly simple alternative of computing separate principal strains and applying them to the fiber directions instead of the real strains would be incorrect in such a case.

Fortunately, a simple physically realistic resolution of this problem is available for the cases of most common interest in which stiff strong fibers are embedded in soft matrices. It is apparent that any shear stresses with respect to the fiber axes can cause neither longitudinal nor transverse stresses between those same axes. The principal effect of such shear stresses is to cause longitudinal and transverse direct stresses in those fibers inclined at 45° to the fibers under consideration. Such an effect is easily accounted for when checking the strength of those

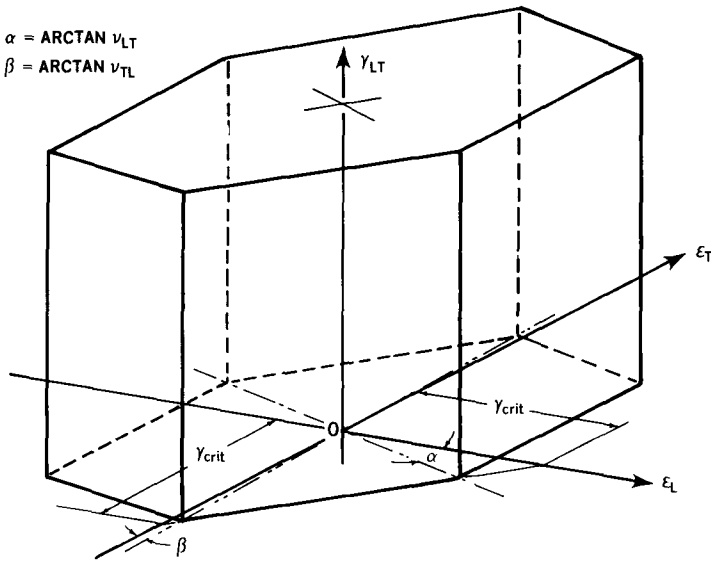


FIG. 17—Strain-based lamina failure criterion.

other fibers under a combination of only axial and transverse loads. Likewise, any shear between axes inclined at $\pm 45^\circ$ with respect to the 0° reference fibers is accounted for in the form of direct longitudinal and transverse stresses in the 0 and 90° fibers.

Nevertheless, the “minor” effect of the shear deformation is of potential concern. If the composite were truly homogeneous, both the fibers and the matrix would undergo the same shear strain in addition to any direct longitudinal and transverse strains. It would not be permissible to ignore the effects of such shear strains on the strength of the fibers. However, when the matrix is much softer than the fibers, most of the shear strain will be confined to the (resin) matrix, and the fibers will not experience significant shear stresses. In such a case, it is reasonable to complete the “lamina” characteristic in the form shown in Fig. 17. The in-plane shear strain cutoff shown for $0^\circ/90^\circ$ shear on the resin matrix would normally be truncated by prior failure of the $\pm 45^\circ$ fibers carrying the shear load. However, there would be no difficulty in applying the matrix shear strain as the dominant limit for very brittle matrices.

Unless there is a compelling reason to do otherwise, it is recommended here that the minor in-plane shear effects be ignored and consideration be given only to the associated direct loads in the $\pm 45^\circ$ fibers and possibly the matrix-dominated $0^\circ/90^\circ$ strength under those shear loads. Materials for which this simplification is reasonable include carbon-, fiberglass-, and boron-reinforced organic matrices like the epoxies.

The author advises caution in applying *any* simplified failure criteria to the more exotic composites under investigation today and reminds readers again that the very useful simple methods presented here should be applied only to conventional fibrous composites such as carbon epoxies, in which strong, stiff fibers are embedded in relatively soft matrices.

Fiberglass-Reinforced Laminates

The case of fiberglass-reinforced plastics may appear to be an exception to the procedures recommended here. The fibers are considered amorphous and incapable of yielding as well as

possibly being isotropic. Nevertheless, when the basic shear-failure mechanism is combined with physically plausible cutoffs for different failure mechanisms, the present theory is indeed able to characterize fiber-dominated failures of such composites.

Consider first the case of a block of glass (left in Fig. 18). There is general agreement that the compression-dominated failures are by shear, just as for metals. The lack of ductility precludes the formation of slip planes, but the failure is otherwise properly characterized by a *difference* between principal stresses. Failure under tensile loads, on the other hand, is known to be dominated by brittle fracture originating at flaws, and the failure envelope is truncated as shown. No other failure mechanisms are needed to encompass the entire failure envelope.

The effect of changing from a block of glass to many fine fibers is to change the ratio of surface area to volume to such an extent that the number of flaws is substantially reduced. Indeed, as suggested at the right of Fig. 18, it may even be possible to raise the threshold for flaw-sensitive failures above the threshold for shear. In addition, the small diameter of the fiber leads to instability failures under compression. It is not unreasonable to postulate that all *three* failure modes could actually coexist at the right of Fig. 18. However, if that were the case, it would probably be necessary to base the “shear-dominated” failures conservatively on the higher of the tensile or compressive strengths because of the lack of a reliable means of measuring the shear strength.

As far as establishing design strengths is concerned, it really does not matter whether or not the nearly vertical and horizontal lines in Fig. 18 actually represent shear failures of the fibers. Either way, they can properly be located through measured tensile and compressive strengths. On the other hand, if the existence of the shear-failure mechanism is denied merely because of difficulty in detecting it under uniaxial loading, one is left with no mechanism for truncating the otherwise rectangular failure envelope for shear-dominated loads. The lack of such truncation is the only significant problem with the popular maximum-strain failure model as a reliable tool for design of many fibrous composite structures.

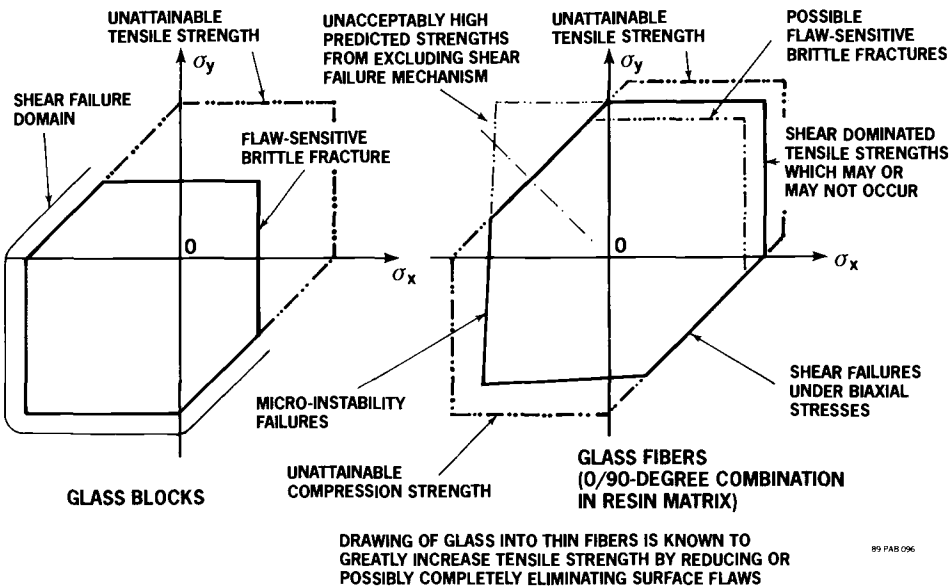


FIG. 18—Explanation of failure mode interaction in terms of glass blocks and fibers.

Experimental Verification

The author's research into predicting the strength of cross-plyed composite laminates was motivated by the development over 15 years ago of a reliable in-plane shear test specimen, shown in Fig. 19. The laminate strengths achieved on that specimen, which were twice as high as had been obtained on more customary contemporary shear test coupons, were still only about half as high as predicted by contemporary failure criteria. Almost another decade passed before the author could determine what was wrong with those theories. With a fuller understanding of the subject, it became apparent that unidirectionally loaded test coupons were so completely dominated by the behavior of fibers aligned in the same direction as the load that they could never invalidate faulty strength-prediction theories. Only biaxial tests were sufficiently sensitive to other properties as well to be able to validate theories. Accordingly, the author prepared a series of papers on biaxial test specimens (see Refs 9 through 11) to make it possible to discriminate between failure theories that were suitable for fibrous composites and those that were not.

Fortunately, other researchers felt the same way. Swanson, in particular, generated biaxial test data on carefully fabricated and tested pressurized axially loaded tubes. The results of Swanson et al. [12] are reproduced in Fig. 20 and compared with the present theory. The agreement is obviously excellent for the tensile-tensile biaxial loads covered by the first series of tests and identified by the solid symbols.

However, the author suspects that the second batch of tubes was either made with more precision or of stronger material since the open symbols display a higher strength at the uni-axial hoop load point. The predicted shear failures, based on the earlier tensile strengths, seem to be premature, but would agree well if the 45° sloping lines were moved outward to pass through the higher unidirectional strain to failure for the later tests (identified by the open symbols). It was obvious to Swanson et al. [12] that the predominantly compressively loaded specimens were failing prematurely, even though they were unable to find reasons for this.

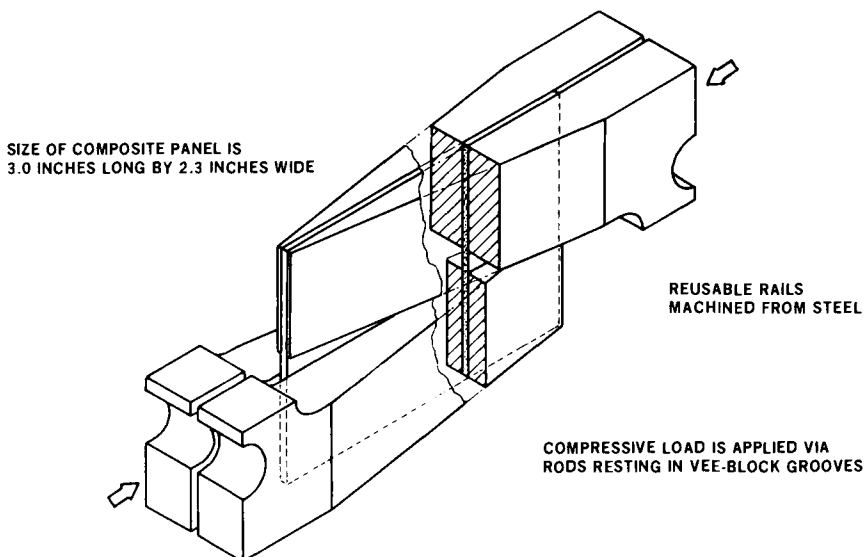


FIG. 19—Douglas bonded-tapered rail-shear test specimen.

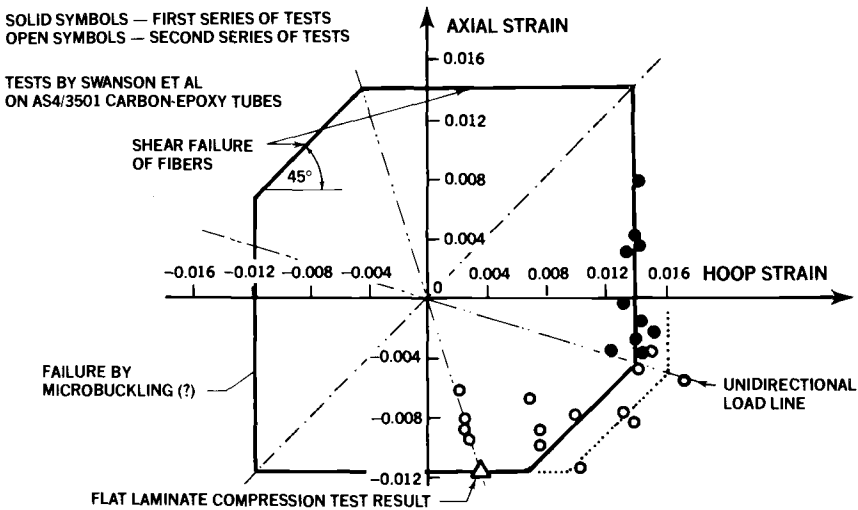


FIG. 20—Comparison of test and theory for truncated maximum-strain failure model for carbon-epoxy composites.

The predicted compressive strengths shown in Fig. 20 are based on the unidirectional compressive strain to failure cited by Swanson et al. [12], but even so, the tubes failed at a still lower strain. Some, but not all, of the most premature failures were associated with compression of the thinnest tubes tested. Even when all these premature failures are explained, it is likely that the cause would recur in real structures subject to the same kind of loading, for example, submersible vehicles.

It is evident that some mechanism other than shear failures of the fibers must have governed those compressive failures. However, as shown in Fig. 15, the present theory makes provision for such a possibility, and it is not at all difficult to add a cutoff representing a different compressive failure mode in part of a predominantly shear-failure envelope. Indeed, Fig. 20 makes such a distinction.

Unfortunately, tests of biaxially loaded composite specimens are often conducted without analysis of probable failure modes and strengths, and most of these data are generally regarded as unreliable. Therefore, such tests failed to influence researchers developing composite failure criteria. And, since the far more numerous uniaxially loaded test specimens were incapable of exposing defective theories, analytical improvements were not forthcoming. Further experimental confirmation of the present theory, subject to the already stated caveats about fiber-dominated failures, may be found in Refs 13 and 14.

A Broader Perspective

As shown in this paper, a failure criterion can be based on a single shear *strain* allowable if hygrothermal strains that do not cause any stresses are eliminated from consideration. Shear strains induced by Poisson contractions in the absence of any shear stress must also be eliminated. There is no counterpart of this phenomenon for isotropic materials that have only one Poisson's ratio. However, for materials having different Poisson's ratios with respect to different material reference axes, the presence of strain in the absence of equivalent stress is quite possible. Heating an unrestrained homogeneous isotropic material cannot create either shear stresses or strains. For an orthotropic material, on the other hand, if the coefficient of thermal

expansion varies with orientation, the principal thermally induced strains will differ so that shear strains must be developed. However, since the orthotropic object is presumed to be unrestrained while being heated, there can be no stresses. In a strict mathematical sense, this phenomenon and the equivalent one involving Poisson contractions are covered by assigning the value zero to the appropriate coefficients in the thermoelastic relations.

In the absence of applied normal stresses, an axial load in a unidirectional lamina will cause the same Poisson contractions in both the transverse and normal directions. Therefore, this load cannot possibly contribute to any shear strains in the plane of isotropy perpendicular to the fiber axes. This is confirmed by rearranging Eqs 7 and 8 for transversely isotropic materials to express the difference between normal and transverse contractions for any combination of axial and transverse loads. It is found that

$$\epsilon_T - \epsilon_N = (1 + \nu_{TN})\sigma_T/E_T \quad (11)$$

no matter what longitudinal stress is applied.

The corresponding principal strain differences between the other pairs of axes follow similarly from Eqs 6 to 8. First,

$$\epsilon_L - \epsilon_N = \left(\frac{1 + \nu_{LT}}{E_L} \right) \sigma_L + \left(\frac{\nu_{TN} - \nu_{TL}}{E_T} \right) \sigma_T \quad (12)$$

in which the second term cannot contribute to the stresses in the L - N plane and must therefore be omitted from the failure criterion. Finally,

$$\epsilon_L - \epsilon_T = \left(\frac{1 + \nu_{LT}}{E_L} \right) \sigma_L - \left(\frac{1 + \nu_{TL}}{E_T} \right) \sigma_T \quad (13)$$

in which *both* terms contribute fully to the stresses in the L - T plane, but the first has no effect on the T - N plane.

With reference to Fig. 15, the 45° sloping lines are defined by both terms in Eq. 13, while the vertical lines are defined by the first term on the right side of Eq. 12, and the horizontal lines are defined by the mirror images of the vertical lines. Equation 11 is assumed here not to be a governing strength limit for strong, stiff fibers in a soft matrix.

What concerns the author most is not good designs in today's composite materials, it is the identification of impractical fiber patterns that lead to weak structures with matrix-dominated failures. Fiber-dominated failures can be analyzed using a theory as simple as the present one. The more complex theories accounting properly for matrix-dominated failures are needed only to identify fiber patterns that should not be used and for which, therefore, there should be no need to predict the strengths. There is also a concern that the present theory is probably inadequate for the more exotic composite materials of the future. It is time to stop pretending that there is a scientific validity to the majority of the methods used to predict the strength of fibrous composite materials no matter how elegant the mathematics, and it is time to acknowledge the need for much new work on a secure foundation. There are opportunities for tremendous contributions to this field, provided that they have the scientific realism needed for such analyses.

Since the presentation of his first papers on composite failure theory [6], the author has been progressively deriving various pieces of the theory needed for a completely rigorous two-phase composite strength-prediction theory. The intent had been that, on its completion, that work would be simplified for the customary fibrous composites in use today and any approxima-

tions made would be justified by comparison with the more general theory. As fate would have it, the second step has been completed first. Nevertheless, a summary of certain features of the more precise analyses may help to guide future researchers.

Figure 21 indicates the features needed to be included in a proper two-constituent (phase) characterization of the strength of cross-plyed fibrous composite laminates. One or the other envelope may dominate, depending on the particular state of stress. And, although the shear-strain plateaus are not shown, the same consideration of one or the other being dominant is just as applicable as for the interaction between the direct in-plane loads.

So that the individual failure envelopes for each constituent (or phase) of the composite material may be superimposed as shown, all strains must be expressed at some common level, for example, that of the lamina rather than at the constituent level. For example, longitudinal strains parallel to fibers are inevitably the same for both fibers and matrix, except for a minute edge zone. On the other hand, the strains perpendicular to the fibers need not be common and are usually not. Therefore, the fiber failure characteristic in Fig. 21 could appear to have a different Poisson's ratio than an isolated fiber, for example. Likewise, for the resin matrix in which the transverse strains vary considerably because of the fibrous "inclusions," the transverse strains in Fig. 21 would have to be averaged over a small but finite distance. The "failure" strain shown would correspond to some different strain at the microscopic level that was actually associated with the stress at which failure occurred.

This difference between constituent and equivalent "lamina" behavior is explained in Fig. 22, where there are two shear failure lines drawn for the fiber failure envelope. The fiber is presumed to have a higher transverse stiffness than the resin matrix, so for a common axial strain under a unidirectional load, two different lateral contractions (corresponding to two different Poisson's ratios) are established. It is, of course, the same failure characteristic, but it can be superimposed on matrix failures in only one form. Figure 22 shows also the corresponding locations of the pure in-plane shear failure locations. It is apparent that, in this case, the blind use of the "lamina" rather than the "fiber" Poisson's ratio would result in an underes-

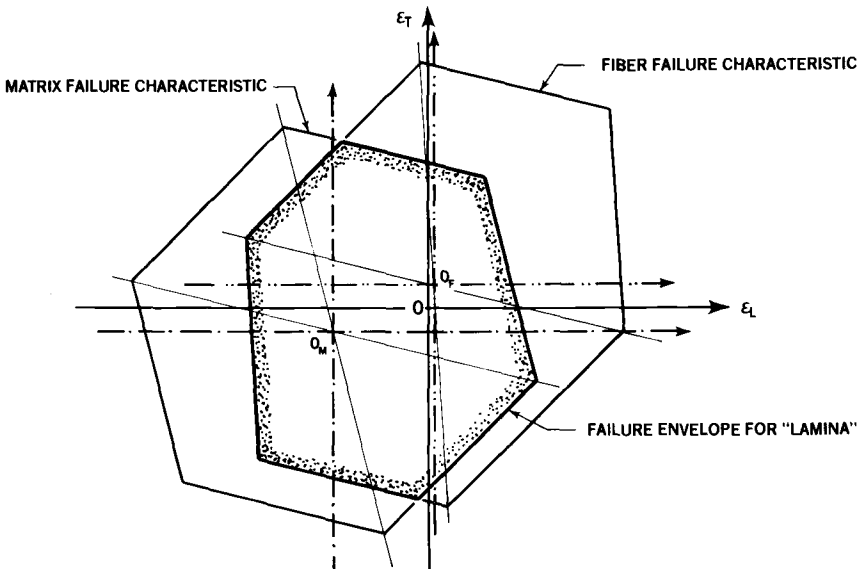


FIG. 21. Separate failure characteristics for fibers and matrix.

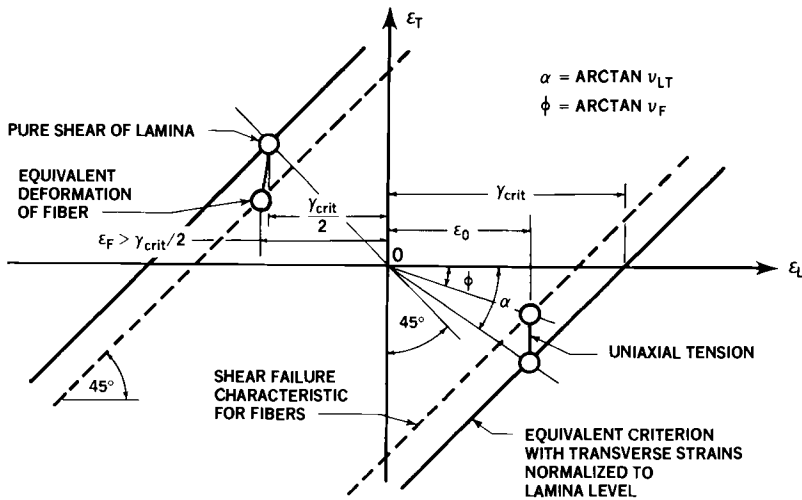


FIG. 22—Equivalent shear failure criteria at lamina and fiber level.

timation of the axial strain in the fiber at which in-plane shear failure occurred. (But the converse could happen for titanium-matrix composites, for example.)

Figure 22 is not only designed to explain how the generalization of the classical maximum-shear-stress failure criterion can still be applied in combination with micromechanics analysis for the more exotic composite materials, but to reinforce the suggestion that there is no need to go to such lengths for conventional fiber/resin composites in which the differences between the Poisson contractions of the fiber and lamina are small enough to be neglected.

The shape of the individual failure envelopes in Fig. 21 for the fibers and matrix could change with the operating temperature. Similarly, there could be a shift in the origins O_F and O_M which are offset appropriately to allow for residual thermal stresses in each constituent from curing at an elevated temperature and operating the structure at high temperatures, where it is very cold. The amount of shift between the origins can be computed on the basis of micromechanical analysis. Since the origin shifts and the matrix failure envelope would vary with operating temperature, the form of the overlapping region of the individual failure envelopes would correspondingly change with service temperature as well. Such an effect cannot be accounted for in any homogenized model supposedly characterizing any composite material.

There is no evidence to suggest that these residual thermal stresses can creep out and relieve themselves significantly. When the residual stress intensity is high at room temperature or below, the matrix is highly resistant to creep. And, when the creep resistance is reduced at elevated temperatures, so are the stresses that would cause the creep. Moreover, the suggestion that the residual thermal stresses can be nullified by swelling stresses as a result of moisture absorption, merely clouds the issue by making the stress-free temperature vary with the manufacturing and service history for each particular structure. In any event, the rate of diffusion is typically so slow that reliance on any absorbed moisture to relieve curing stresses would in turn introduce the problem of how to treat the effects of that water during a rapid change in temperature.

Note that the residual stresses are definitely *not* a lamina property. Most prior treatments of residual thermal stresses have started with the false premise that the individual lamina is homogeneous and inherently stress-free in isolation, no matter what its temperature. Such an

oversimplification cannot possibly explain matrix cracking. It is even questionable for use in characterizing edge delaminations resulting from stacking together too many parallel layers of unidirectional material, whether these delaminations be caused by thermal or mechanical stresses.

Matrix failures, either intraply or interply, *always* represent an inability to develop the full strength of the fibers and should therefore be avoided by choice of an appropriate fiber pattern and layer sequence for each laminate. Abstract mathematical composite failure theories have not adequately addressed matrix failures and, until they do, one should use good judgment to exclude the inferior arrangements rather than believe that such constraints do not exist merely because one's computer program is not smart enough to advise users of its weaknesses. Actually, far more restrictive limits on fiber patterns arise when one considers bolt holes, which introduce even further matrix-dominated influences on the degree of stress-concentration relief. The author believes that matrix-dominated failures will continue to require empiricism even after rigorous two-phase theories have been developed for unnotched laminates. So, the use of a simple failure model that adequately characterizes only fiber-dominated failures can be just as reliable as more elaborate theories, provided that it is used in conjunction with realistic constraints on fiber patterns.

Implementation of Maximum-Shear-Stress Failure Criterion for Fibrous Composite Laminates

Implementation of this new method for predicting the fiber-dominated strengths of composite laminates is remarkably simple and requires far *less* input data than older failure models. Figure 23 shows the step-by-step construction of the failure envelope for the case in which the tensile strength exceeds the compressive strength. (Similar envelopes could be constructed for other cases.) The uniaxial tensile and compressive strengths are established as in the diagram at the left of the figure. The middle diagram shows how the envelope is built around those points, with vertical lines for compressive instability failures and lines at slopes defined by ν_{LT}

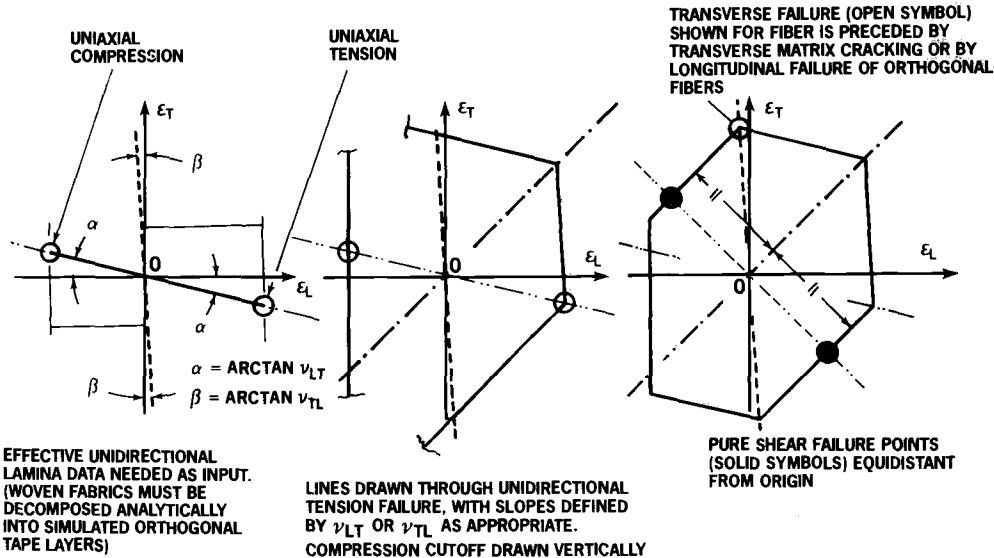


FIG. 23 Construction of failure envelopes for fiber-dominated laminate strengths.

and ν_{TL} and lines at 45° for the shear-dominated failures. The final diagram, on the right, lacks most of the symmetry in Fig. 8 because of the difference between tensile and compressive strengths. But, a combination of Fig. 23 with its own mirror image whenever appropriate would make it seem similar to the left diagram in Fig. 15.

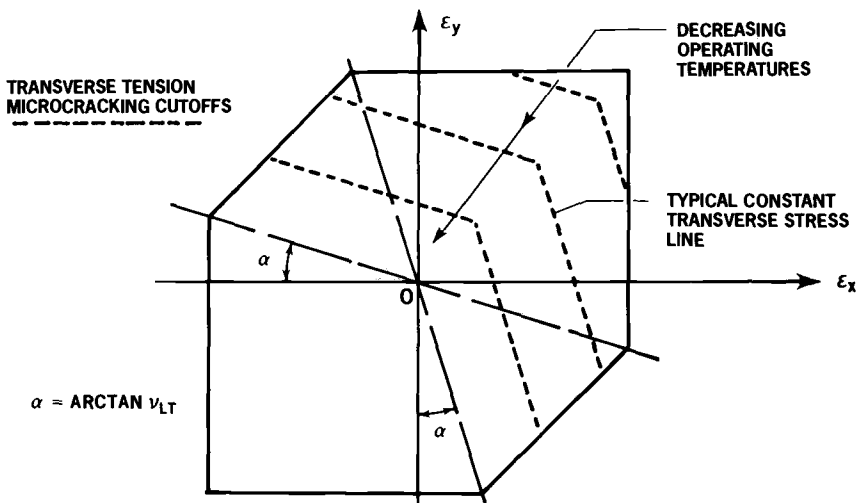
The reliability of this method of analysis is enhanced tremendously by eliminating any opportunity to distort the failure envelope by the transverse-tensile strength—the most unreliable of all the measurements used for earlier theories.

In those few instances when noncatastrophic transverse-tensile failures actually do occur, the present theory provides a framework for including the effect. Figure 24 shows how this would be done for the combination of a high-strain fiber such as glass with a very brittle matrix. The figure has been prepared for a *combination* of orthogonal fiber directions because the primary effect of transverse cracking of a 90° ply is to impose a reduction on the allowable axial strain in the 0° ply—it is not merely a problem for the transverse ply itself.

There is a very real problem in characterizing this “failure” mode, regardless of the model being used. The application of a subcritical tensile load at very low temperature might well crack the matrix without failing the fibers. Then, if a major compressive load were subsequently applied at a higher temperature, there would be no warning that the compressive strength shown could not be attained. Similarly, the less severe truncations shown for higher operating temperatures would be invalid if the matrix had already been cracked extensively by an earlier cold load cycle.

This new composite failure model has been criticized for omitting consideration of matrix-dominated failures and thus being incomplete. However, other theories without such stated restrictions cannot calculate matrix-dominated strengths either. The difference is not in the capabilities of the theories in this regard, but in the misuses to which they may inadvertently be put.

Actually, belief in such calculated matrix-dominated strengths has created enormous problems. It was bad enough trying to predict the progressive failure of a test coupon under a single



NOTE: NO CHANGE TO PREDICTED FAILURES FOR OTHER STATES OF STRESS

FIG. 24. Physically realistic treatment of matrix cracking occurring before fiber failures.

load condition. Yet the design and analysis of complete composite structures that were subject to multiple loads in a random order would be chaotic if structurally significant matrix failures were tolerated. It is already common practice to reduce fiber-dominated in-plane design strain allowables to tolerate specified sizes of delaminations that would be too difficult to detect in service. (Such strain limitations are frequently *more* severe than those imposed because of loaded bolt holes.) Imagine how much worse the inspection problem is for mere interfiber cracking. All failures would have to be defined as catastrophic at what had previously been referred to as only “first-ply” failures. And even those would have to be limited by arranging the various load conditions in the most adverse sequence. Analysis of matrix-dominated failures for any other purpose than to identify laminates that should *not* be used in practice is akin to opening Pandora’s box. It would transform the certification of composite aircraft from a task of great difficulty to one of utter impossibility. The author respectfully submits that the omission of a means for calculating matrix-dominated strengths from this theory is not a liability but an incalculable asset.³

Conclusions

A simple and physically realistic method for predicting the strength of cross-ply fibrous composite laminates has been shown here to follow from a generalization of the classical maximum-shear-stress yield criterion for ductile metal alloys. The generalization to orthotropic materials has required that the failure criterion be expressed in terms of strains rather than the more customary stresses, even though the actual governing criterion is one of stress.

The new failure criterion is shown to be similar to a simple truncation in the in-plane shear quadrants of the well-known maximum-strain empirical failure model for composites. Therefore, this new work is closely related to the best-known classical failure models for both ductile metals and composite laminates. The new theory has also been compared favorably with biaxial test data of other researchers. While the new theory is unlikely to be improved upon for state-of-the-art fiber/polymer composites, there remains a need for a proper two-phase strength analysis. Separate accounting for failures in the fiber or in the matrix will be needed both for matrix-dominated failures in impractical fiber patterns with today’s composites and for predicting any failures of the more exotic composites that are as yet only in the research and early applications stage.

Acknowledgment

Two of the author’s colleagues at Douglas have contributed significantly to this work. Dave Peterson coded the BLACKART computer program used to assess the effects of all the usual input on the output from composite strength-prediction methods. Benson Black was the first to recognize that the failure of *all* cross-ply laminates could be reduced to a *single* shear-failure characteristic regardless of the fiber pattern. Even without the benefit of this latest research, the empirical shear cutoffs he applied to the use of the maximum-strain failure criterion for C-17 composite components are remarkably close to the best recommendations that could be made today. The information presented here would have taken much longer to develop without such support and advice.

³This statement should not be misinterpreted as a criticism of research into matrix-dominated failures of composite laminates. The limitations on fiber patterns to ensure fiber-dominated failures can be identified only by experience or proper analyses of matrix failures. The author’s concern is not about such analyses themselves, but rather about the many misuses to which they will inevitably be put by mathematicians and engineers who have failed to understand even the phenomena associated with micromechanical behavior of fibrous composites.

The author would also like to acknowledge the encouragement from the MIL-HDBK-17 committee members and, in particular, the U S Army MTL and FAA representatives to persevere with this work

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Session IV

Advanced Materials Analysis and Test

Robert B. Croman, Chairman

Overview

Session IV is comprised of four papers concerning the performance of fiber-reinforced metal, intermetallic, or ceramic matrices subjected to mechanical loading. Two of the papers address the material's performance when it is subjected to fatigue loading at an elevated temperature. The other two papers detail the material's performance when it is subjected to statically applied loads. All four papers spend considerable time detailing the different aspects of loading or failure history or both.

Pollock and Johnson present the paper, "Characterization of Unnotched SCS-6/Ti-15-3 Metal Matrix Composites at 650°C," and in it predict the laminate static stress-strain response and fatigue cyclic stress-strain behavior from the constituent material behavior at elevated temperature. A number of different fiber layups are considered. It is found that the SCS-6 silicon-carbide fibers reinforcing the Ti-15-3 alloy metal matrix in the loading direction are the most influential component contributing to elastic modulus, tensile strength, and elongation to failure. Furthermore, they find a direct correlation between fatigue endurance and the percentage of fibers in the loading direction. Furthermore, they conclude that the fibers contribute more to the fatigue life performance at elevated temperatures rather than at room temperature conditions because of the greater load carrying of the fibers relative to the lower modulus matrix. However, inhibiting this added role of the fibers to fatigue at elevated temperature, is an observation that the in-situ fiber strength decreases by 20 to 30% at elevated temperatures, which is not the case for single fibers tested at elevated temperatures. Looking at fatigue-induced damage, the authors conclude that high fatigue loads that result in short-lived specimens show only multiple fiber breaks and no matrix failure, whereas at low fatigue loads and, hence, long life, the specimens showed extensive matrix cracking and no fiber fractures outside of the immediate vicinity of the matrix cracks. The authors also observe that during the fatigue-loading process the initial unloading modulus is higher than the loading modulus, and they attribute this to a matrix creep behavior during the unloading process.

Bartollota and Brindley also investigate the fatigue behavior of an advanced composite system in the paper, "High Temperature Fatigue Behavior of a SiC/Ti-24Al-11Nb Composite." Here, the matrix Ti-24Al-11Nb was an intermetallic and reinforced with SiC fiber in which the fiber volume loading is $\sim 27\%$, and the fiber alignment is unidirectional. Fatigue testing is performed at 425 and 815°C. Fatigue testing is performed in tension using either load- or strain-controlled conditions. All the data are analyzed by plotting the maximum fatigue strain versus the number of cycles to failure. These plots show three regimes of performance. Interestingly, the high load, low life, Region I and the low load, high life, Region III are not influenced by test temperature, but the intermediate Region II shows definite temperature-dependent performance. Region I shows a catastrophic failure performance which is mainly defined by the failure of the fiber. Region II shows a progressive failure behavior, in which in the strain-controlled tests, the load at first gradually decreases then abruptly decreases to final failure. The Region III conditions produce runout and define an endurance limit. The authors further discuss the merits and disadvantages of load- and strain-controlled testing in fatigue and conclude that the use of both is complementary and help define the failure process.

Lee and Daniel use a modified shear lag analysis to describe the tensile static stress-strain behavior of a unidirectional ceramic matrix composite in their paper, "Deformation and Fail-

ure of Longitudinally Loaded Brittle-Matrix Composites” The analysis uses a single fiber embedded in a cylindrical matrix model The matrix is assumed to have lower strain to failure than the fiber, therefore, the matrix is assumed to fail by transverse cracking before any fiber failure The shear stress in the fiber is assumed to be a linear function of the radial coordinate variable, and the shear stress in the matrix is assumed to be an inverse function of the radial coordinate variable to the second power The authors consider three cases (1) an embedded fiber of finite length in an infinite matrix medium, (2) an embedded fiber of infinite length in a semi-infinite matrix medium, and (3) a fiber in a cylindrical matrix with two transverse matrix cracks. By using various combinations of Cases 2 and 3, the authors are able to predict the stress-strain behavior and stress-versus-matrix crack density of a silicon-carbide unidirectional reinforced glass-ceramic matrix composite subject to longitudinal tensile loading The predictions are found to be in reasonably good agreement with the experimental results. In addition, the prediction accurately determines the knee of the stress-strain curve for the material considered, although the subsequent secondary loading is predicted at a somewhat lower value of strain than the experimental results show.

The last paper in this session, “A Macro-Micromechanics Analysis of a Notched Metal Matrix Composite” by Bigelow and Naik, is a numerical analysis of a unidirectional composite with central notch subject to tensile loading To demonstrate the utility of the approach, a boron-reinforced aluminum matrix monolayer having a fiber volume loading of 30% is considered The overall composite is analyzed by a three-dimensional finite element program called PAFAC The element in the vicinity of the central notch is further analyzed on a micro level by a discrete fiber-matrix (DFM) model which uses as its boundary condition the results from the three-dimensional finite element analysis of the macro scale This combined analysis shows some aspects of the stress distribution in the vicinity of the notch For instance, a unit value of applied stress to the sample results in a distribution of axial stress with maximum value of 32 MPa in the fiber closest to the notch Furthermore, the combined analysis predicts that yielding in the matrix near the notch will occur when the remote stress is as low as 5.75 MPa. This is considerably lower than the 8-MPa value that will be predicted by a macro level analysis by itself

In closing, these four session papers illustrate that much has been done in understanding the new composite systems using metal, intermetallic, and ceramic matrices, but much research remains to be done to understand further these materials

Characterization of Unnotched SCS-6/Ti-15-3 Metal Matrix Composites at 650°C

REFERENCE: Pollock, W. D. and Johnson, W. S., "Characterization of Unnotched SCS-6/Ti-15-3 Metal Matrix Composites at 650°C," *Composite Materials: Testing and Design (Tenth Volume)*, ASTM STP 1120, Glenn C. Grimes, Ed., American Society for Testing and Materials, Philadelphia, 1992, pp. 175–191

ABSTRACT: Ti-15-3 reinforced with SCS-6 silicon carbide fibers, in five different lay-ups, was tested at 650°C to determine monotonic and fatigue strengths, basic mechanical properties, and damage initiation and progression. The elevated temperature results were compared to those obtained at room temperature. Analytical predictions were made of the monotonic stress-strain response as well as cyclic stress-strain hysteresis. The fiber reinforcement was found to increase significantly the static and fatigue strengths of the laminates over those of the matrix material at elevated temperature; the increase was insignificant at room temperature. Initial damage, in either the fibers or the matrix, was partitioned as a function of the life and applied strain range in the constituents. High strains and short lives resulted in multiple fiber failure with no signs of matrix fatigue cracking. Low strains and long lives resulted in extensive matrix cracking and no fiber breaks away from the fracture surface. At 650°C, the matrix was too weak to cause fiber-matrix interface failure before matrix yielding. Laminate fatigue lives were hypothesized to be a function of the 0° fiber stress. More scatter was found in the 0° fiber stress versus high temperature fatigue life data than in the room temperature data. An initial unloading modulus that was greater than the initial loading modulus was observed in the elevated temperature fatigue tests.

KEYWORDS: titanium, fatigue, moduli, elevated temperature, silicon-carbide fibers, creep

New aerospace vehicles are being used in conditions that push the capabilities of current materials to their limit. Proposed supersonic and hypersonic vehicles require lightweight materials with good mechanical properties at elevated temperatures. Titanium has better mechanical properties at elevated temperature than aluminum, and although not as light, it is much lighter than most other candidate materials. To improve the strength and stiffness at elevated temperatures and keep the weight low, reinforcement has been added to the titanium. The most common reinforcement is silicon carbide (SiC) fiber. The system used in this research was the Ti-15-3 (15 w/% vanadium, 3 w/% aluminum, 3 w/% chromium, and 3 w/% tin [1]) matrix reinforced with continuous SCS-6 SiC fibers.

The objectives of this research were to gain an understanding of the response of the composite to applied loads at elevated temperature, to compare this behavior to previously obtained room temperature results [2], and to evaluate an analytical technique for predicting composite behavior at elevated temperature using constituent properties. The response of the composite to monotonic and cyclic loading at elevated temperature was evaluated. Curves of maximum stress against fatigue life were generated for several lay-ups. Analytical methods were used to predict the response of the composite to various loading histories using the con-

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stituent properties. Damage initiation and propagation were studied and related to the microstructure and load history.

Material

The SiC fibers, designated SCS-6 by the manufacturer (Textron Specialty Materials), are 0.145 mm in diameter with an elastic modulus of 412 GPa and an ultimate tensile strength of 3.5 to 4.4 GPa [3]. The fibers were laid up between the matrix foils and consolidated by hot pressing. Four different 8-ply lay-ups were tested: $[0]_8$, $[0/90]_{2s}$, $[0_2/\pm 45]_s$, $[0/\pm 45/90]_s$. In addition, a 13-ply panel was made of foils only and processed the same as the composite panels. This material, referred to as foil matrix material, provided an unreinforced matrix subjected to the same processing as the composite. Thicker Ti-15-3 matrix material was also acquired from the same vendor as the foil (Timet). The thicker material is referred to as plate matrix material. The composite specimens and foil matrix specimens tested at elevated temperature were straight-sided, 0.20 cm thick, 15.2 cm long, and 1.3 cm wide. The plate matrix specimens were 0.5 cm thick, 1.3 cm wide, and 15.2 cm long.

Strain-controlled tests simulated the behavior of the matrix material in the composite. In the strain-controlled cyclic tests, once the maximum strain was achieved, the matrix stress dropped rapidly as a result of time-dependent deformation. At the minimum strain, compressive loads were required to compensate for the time-dependent deformation. These compressive loads caused the thin foil matrix coupons to buckle. Fixtures to prevent buckling could not be used at the elevated temperature. Therefore, the thicker plates of matrix were needed for the strain-controlled cyclic tests.

Test Procedures

All tests were conducted in a servohydraulic test frame. The grips were water-cooled, and the samples were heated by a 5-kW induction heater. The temperature was monitored by two infrared pyrometers, and a microprocessor controlled the induction heater. All cyclic tests were performed at 10 Hz with an R ratio ($P_{\min}/P_{\max}$) of 0.1. The load-controlled monotonic tests were loaded at a rate of 0.273 kN/s. This rate was chosen to match the rate used in the room temperature tests [2]. The strain-controlled monotonic test strain rate was 2×10^{-4} mm/mm per second. This strain rate matched the load-controlled strain rate of the unidirectional composite test at 650°C. The strain was monitored by a water-cooled, high-temperature extensometer with two quartz rods that contacted the coupon.

An infrared camera was used to monitor the thermal gradient across the test section. The infrared image provides a full-field view of the sample as opposed to a thermocouple point reference. With this monitoring method, the thermal gradient at 650°C was reduced to 10°C over a 50-mm length. This length spanned the 25-mm gage length. The maximum load and strain used during the cyclic tests are presented in Table 1. Because of the shortage of material, only one test was conducted for each condition. During the cyclic tests, the loading rate of some individual cycles was slowed to 0.273 kN/s so the stress-strain curve could be plotted on an $x-y$ recorder. Typically, the stress-strain curve for Cycles 1 through 5, 10, 20, 50, 100, 500, 1000, and so on were recorded.

Test Temperature and Hold Time

A test temperature of 650°C was selected for two reasons. First, this temperature appears to be the highest use temperature considered for this composite. Second, preliminary tests indicated that the matrix material exhibited time-dependent deformation at this temperature. These tests are described in the next section.

TABLE 1—650°C fatigue test results

Lay-Up	Max Stress (Strain) ^a	Percent of Ultimate Tensile Strength	Cycles to Failure	Type of Control
Matrix plate	214 MPa	80	4	load
	172 MPa	65	150	load
	131 MPa	50	37 000	load
	92 MPa	35	200 000	load
	0 0088 mm/mm		2 890	strain
	0 0077 mm/mm		13 300	strain
	0 0047 mm/mm		109 000	strain
[0] ₈	896 MPa	95	20 000	load
	848 MPa	90	44 000	load
	759 MPa	80	183 000	load
[0/90] _{2s}	493 MPa	95	100	load
	466 MPa	85	7 800	load
	383 MPa	70	40 000	load
[0/±45/90] _s	379 MPa	90	50	load
	337 MPa	80	64 000	load

^a*R* = 0.1 for load and strain control

Infrared pyrometers provided the temperature feedback to the controller. This feedback depends on the surface emissivity of the titanium. The emissivity changes with time at temperature as oxides form on the surface. A few tests, using a thermocouple as a reference for calibration, showed that at 650°C the oxide formation and, therefore, the emissivity stabilized after 30 min. Therefore, a hold time of at least 30 min at 650°C was required before testing to ensure stable temperature readings.

This metastable beta alloy exhibits at elevated temperatures a phase change that can affect the mechanical properties [4]. Elevated temperatures also cause reactions at the SiC fiber-titanium matrix interface that can change the response of the composite. There was some concern that these metallurgical changes would cause significant scatter in the experimental results. Differential scanning calorimetry (DSC) tests were conducted to study the stability of the composite at 650°C. DSC measures the heat flow into a sample as a function of time and temperature. The sample was heated to 650°C and held at this temperature for 150 min. Because heat flow is required for metallurgical reactions such as phase changes or interfacial reactions, this test indicates how long it takes a sample to stabilize during an isothermal hold.

Analytical Model

One of the objectives of this research was to evaluate an analytical technique for predicting the mechanical behavior of the titanium matrix composites at elevated temperature using constituent properties. The laminate stress-strain curves to failure were predicted using the AGLPLY program [5], the same program used to predict the room temperature mechanical properties [2]. This program performs an elastic-plastic analysis of composite laminated plates under inplane mechanical loads. The program uses the vanishing-fiber-diameter (VFD) model [5] to predict laminate behavior. The VFD model is a homogeneous, orthotropic material model based on constituent properties. The fiber and matrix properties are input for the analysis. The experimentally observed elastic-plastic stress-strain curves for the matrix material were used with no empirical correction factors. AGLPLY does not predict time-dependent deformation. The model assumes perfect fiber/matrix bonding. Thus, if fiber/matrix interfaces fail, this program would predict a somewhat stiffer response than would be experimentally observed. The AGLPLY program computes the overall laminate elastic moduli, the aver-

age stresses and strains in the fiber and matrix of each ply, and the overall strains of the composite for the entire loading regime. Laminate properties predicted by the AGLPLY program compared favorably with laminate properties predicted by other material models developed for metal matrix composites [6].

Results and Discussion

This section describes the microstructural analysis and compares experimental and predicted results. First, the data used to design the experiments and a study of the microstructure of the composite are presented. Then the mechanical properties of the matrix and both experimental and predicted mechanical properties of the composite are discussed. Last, a fractographic study of the matrix and composite after monotonic and cyclic loading are discussed.

Preliminary tests were performed on the matrix to determine the minimum temperature at which time-dependent deformation is significant. A constant stress of 70 MPa was applied, and the temperature was increased incrementally while the strain was monitored. At each temperature level, the temperature was held for 1 h. Although some time-dependent deformation was observed at 480°C, it was not significant. At 650°C, however, there was a significant amount of time-dependent deformation; in fact, the matrix material continued to deform after the hold period of 1 h at 650°C. Therefore, at 650°C, time-dependent deformation occurs in the matrix and must be properly accounted for in the analytical predictions.

DSC analysis was done to study the thermal stability of the composite at 650°C. The heat flow, measured by the DSC, increased rapidly during the first 50 min of the 650°C isothermal hold and then approached a constant value (that is, the heat flow rate approached zero). Therefore, in all tests, an initial 50-min hold time was used. This resulted in less scatter in the data due to the relatively low rate at which metallurgical reactions occurred. This initial hold was also more than sufficient to build an oxide layer for stable emissivity readings.

Composite Microstructure

The microstructure of the as-received $[0]_8$ composite is shown in Fig. 1a. The carbon core of the fibers is visible as well as the carbon-rich coating on the fibers. Many of the grains in the matrix material extend from fiber row to fiber row. These grains consist mainly of metastable beta phase, a body-centered cubic (bcc) crystal. Brittle titanium carbides can form along the prior foil boundaries as a result of poor processing [7]. Figure 1b shows the microstructure of a sample held at 650°C for 1 h. Small amounts of the alpha phase, an hexagonal-close packed (hcp) crystal, formed during this hold time at temperature. The alpha forms along the grain and subgrain boundaries. The alpha typically has improved tensile strength over the beta, but significantly reduces elongation to failure [4].

Monotonic Response

This section describes the stress-strain response of the matrix and the composite to monotonic loading at both room and elevated temperature. Results from load- and strain-controlled tests are compared for the matrix. The monotonic response of the composite was predicted using the monotonic response of the matrix.

Matrix Material—At both room temperature and 650°C, significant differences were observed in the stress-strain response of the matrix material when testing in load or strain control.

At room temperature, the modulus was independent of the control mode. As shown in Table 2, the ultimate tensile strength from the strain-controlled test was 19% lower than from

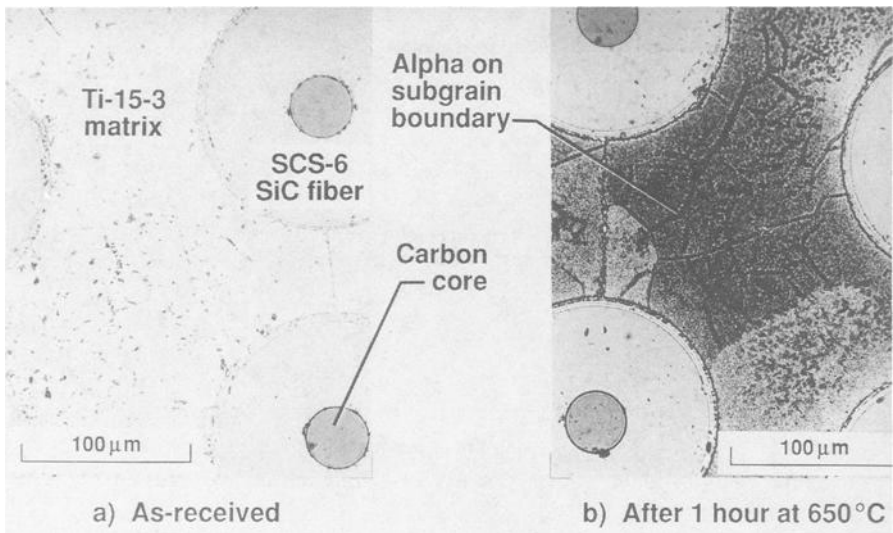


FIG. 1—Microstructure of the SCS-6/Ti-15-3 composite. Both a and b were etched with Kroll's.

the load-controlled test. Presumably, the difference in strength is due to the difference in strain rate after yield. The strain to failure is presented with these results in Table 2.

Figure 2 shows the stress-strain curve of the matrix at 650°C in load and strain control. As at room temperature, the elastic modulus at 650°C was independent of the control mode, and the inelastic response of the matrix depended on the control mode. The stress level in the strain-controlled test reached 145 MPa but did not work harden (that is, the load dropped and plastic deformation continued at a stress below 145 MPa). At this loading rate, a portion of the deformation was time dependent. Under load control, the material did exhibit work hardening and failed at an ultimate tensile strength of 335 MPa. The strain to failure in both tests

TABLE 2—Mechanical properties of the matrix and the composite at room temperature and 650°C.

Lay-Up	Test Temperature	Ultimate Tensile Strength, MPa	Initial Elastic Modulus, GPa	Strain to Failure, mm/mm	Type of Control
Matrix	RT	934	92		load
Matrix	RT	754	88		strain
Matrix	650°C	335	76	>0.150	load
Matrix	650°C	145	77	>0.150	strain
[0] ₈	RT	1517	189	0.009	load
[0] ₈	650°C	948	178	0.0077	load
Predicted	650°C	948 ^a	181	0.0066	load
[0/90] _{2s}	RT	945	138	0.0097	load
[0/90] _{2s}	650°C	548	126	0.0070	load
Predicted	650°C	561	154	0.0065	load
[0 ₂ /±45] _s	RT	1069	156	0.0105	load
[0 ₂ /±45] _s	650°C	554	125	0.0070	load
Predicted	650°C	565	146	0.0066	load
[0/±45/90] _s	RT	752	134	0.0098	load
[0/±45/90] _s	650°C	421	131	0.0077	load
Predicted	650°C	411	129	0.0066	load

^aPrediction fit to data to establish ultimate fiber strength = 2630 MPa.

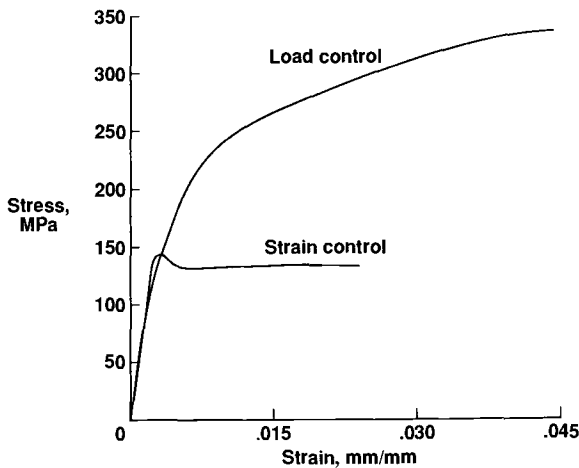


FIG. 2—Stress-strain response of the matrix in strain and load control at 650°C.

exceeded the range of the extensometer, 0.15 mm/mm. The strain-controlled tensile strength at 650°C was 81% lower than at room temperature; for the load-controlled test, the tensile strength dropped only 64%. The differences in behavior for load and strain control illustrate the dependence of the matrix on the loading history.

Composite—This section describes the monotonic response of the composite at room and elevated temperatures. The room temperature [2] and 650°C data, the initial elastic modulus, strain to failure, and ultimate tensile strength for each lay-up are given in Table 2.

For all lay-ups tested, the initial elastic modulus at elevated temperature was 15 to 22% lower than the initial elastic modulus at room temperature. The amount of drop in initial elastic modulus as a result of elevated temperature was primarily a function of the percentage of fibers in the loading or 0° direction. However, the angle of the remaining plies also affected how much the initial elastic modulus decreased because of elevated temperature. The 90° fibers have essentially no length along the loading axis and, therefore, contribute little to the modulus of the composite. The 45° fibers are not in the loading direction but do contribute to the elastic modulus of the composite. The load transferred to the 45° fibers is a function of the yield strength and modulus of the matrix material. The modulus of lay-ups with 45° fibers was more sensitive to a temperature increase because of the reduction in the matrix yield strength and modulus. For instance, both the $[0_2/\pm 45]_s$ and the $[0/90]_{2s}$ have the same percentage of fibers in the loading direction, but the $[0_2/\pm 45]_s$ had the larger drop in initial elastic modulus as a result of elevated temperature.

As was the case for the modulus, the strengths at elevated temperature were lower than the strengths at room temperature for all lay-ups. The decrease in strength is primarily a function of the percentage of fibers in the loading direction.

In general, the data in Table 2 show that at room temperature only the unidirectional laminate has a tensile strength significantly greater than the matrix material alone. However, at 650°C, the yield strength of the matrix material was so low that all of the laminate strengths were significantly greater than the matrix alone. This is one reason that fiber reinforcement is needed for high-temperature applications.

Bhatt [8] showed little degradation of the elastic modulus or ultimate tensile strength of the SCS-6 fiber up to 1100°C. The strains to failure of all the composites at elevated temperature were less than at room temperature, as shown in Table 2. The strain to failure can be multiplied by the fiber modulus to estimate the static strength of the 0° fibers. Therefore, assuming the

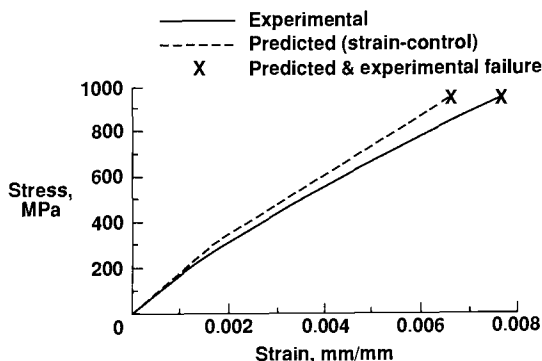


FIG. 3—Unidirectional stress-strain response at elevated temperature.

fiber modulus was constant with temperature, the in-situ strength of the fibers was reduced because of exposure to elevated temperature. The strengths reported by Bhatt [8] were from tests done on single fibers (that is, the fiber not in contact with matrix). Apparently, reactions between the fiber and matrix at elevated temperature lowered the strength of the in-situ fiber.

At room temperature, interfacial failure was seen as a knee in the elastic stress-strain curve [2] occurring well below the yield strength of the matrix material. However, this knee was not observed at elevated temperature. At 650°C, the matrix yielded before loading the interface to failure. Thus, when the interface failed, the knee was obscured because of the plastic deformation in the matrix material. Previous work [2,9] indicated that, with thermal aging, the interface strength may increase over that of the as-fabricated composite. Therefore, the interface strength may have been greater as a result of the 650°C conditioning of the specimen for 1 h before testing.

Analytical Predictions—The experimental and predicted stress-strain behavior of each lay-up at 650°C is presented in Figs. 3 through 6. The predictions were made using the AGLPLY computer program based on the strain-controlled matrix response. Because the predictions made with the strain-controlled matrix material response were in better agreement with the experimental results, only the predictions made from the strain-controlled matrix material response are shown.

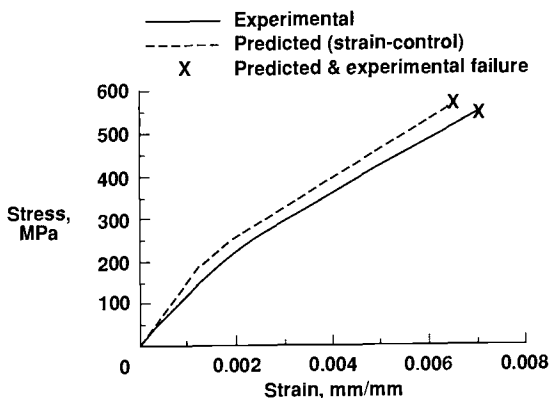


FIG. 4—The [0/90] stress-strain response at elevated temperature.

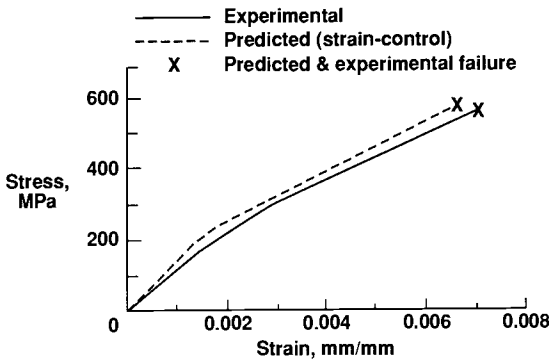


FIG. 5—The $[0_2/\pm 45]_s$ stress-strain response at elevated temperature.

The tensile strength of each lay-up at elevated temperature was also predicted using AGLPLY. The failure criterion was defined from the results of the unidirectional test as follows. The AGLPLY program calculated the fiber stress in the $[0]_8$ to be 2630 MPa at an applied load of 948 MPa, the observed failure load of the unidirectional composite at 650°C. Failure was defined as the applied stress required to load the 0° fibers to a stress of 2630 MPa. By definition, the predicted failure stress of the unidirectional composite was 948 MPa. The stress in the 0° fiber was used as a failure criterion rather than the strain to failure so that the resulting laminate strength predictions would be conservative (approximating the correct strength but lower strains to failure). Alternatively, using correct failure strain would have predicted high strengths. The predicted and experimental failure stresses are also shown in Figs. 3 through 6.

Composite Cyclic Response

Figure 7 shows the cycles to failure as a function of maximum stress for the various lay-ups at 650°C. The data for the $[0_2/\pm 45]_s$ are not shown; there were a large number of defects in this panel, which resulted in large discrepancies in the data. For a given stress level, the unidirectional composite had the longest life, the $[0/90]_{2s}$ was second, and the $[0/\pm 45/90]_s$ composite had the shortest life. Clearly, the higher the percentage of fibers in the loading direction, the more fatigue resistant is the composite.

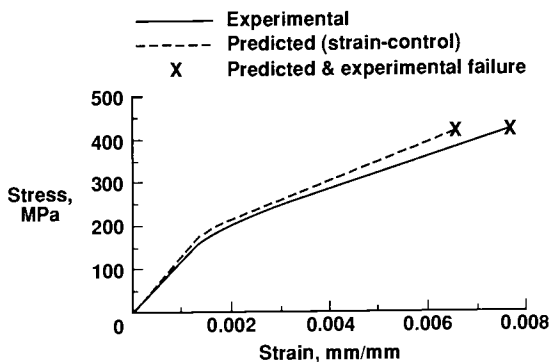


FIG. 6—The $[0/\pm 45/90]_s$ stress-strain response at elevated temperature.

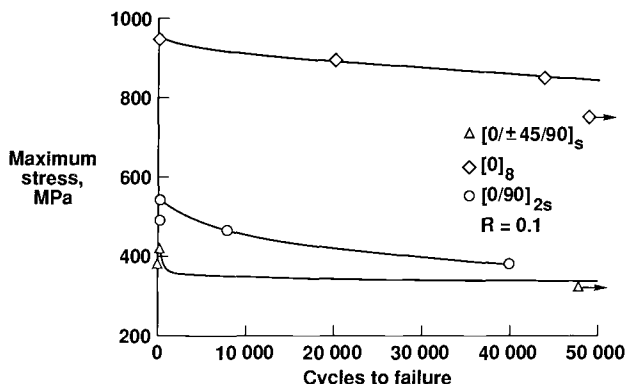


FIG. 7—Cycles to failure for the SCS-6/Ti-15-3 composite at 650°C.

Figure 8 compares the maximum strain versus cycles to failure of the unidirectional composite to the maximum strain versus cycles to failure of the matrix loaded in strain control. The fatigue test of the unidirectional specimen was essentially an in-situ fatigue test of the fibers. For high maximum strains, short lives, the initial damage developed in the fibers and the composite had lower life expectation than the matrix. For low maximum strains, long lives, the initial damage developed in the matrix, and the matrix had lower life expectation than the composite. Where the two curves intersect, both the matrix and the fibers were equally likely to develop fatigue damage.

Figure 9 shows typical initial fatigue damage that developed in the unidirectional specimen for short, medium, and long lives. As shown in Fig. 8, for the unidirectional composite, it is possible to partition the damage into fiber- or matrix-dominated regions using the cyclic life of the composite and the matrix. During the cyclic tests at 650°C, the strain range decreased during the first ten or so cycles. This was opposite to what was reported for the room temperature tests [2]. The strain range was then constant for the remainder of the life. The constant, or stabilized, strain range was multiplied by the modulus of the fiber to obtain the maximum cyclic stress in the 0° fibers. This fiber stress is plotted against life for each lay-up in Fig. 10. Note that the room temperature data [2] fall within a narrow band. While the 650°C data were

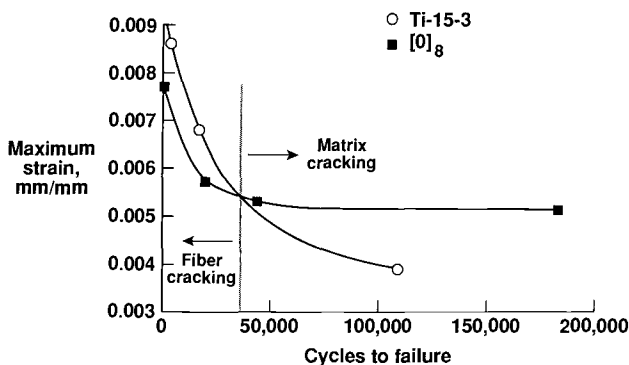


FIG. 8—Life of the unidirectional composite and the matrix as a function of maximum strain at 650°C.

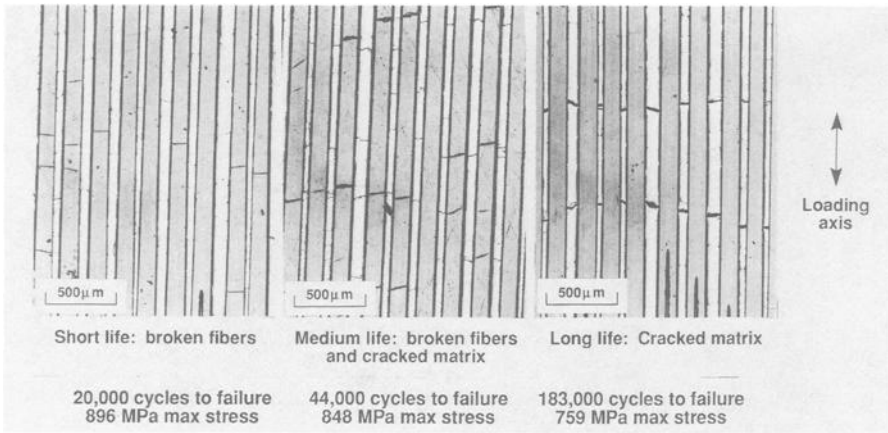


FIG. 9—Sections parallel to the loading axis in unidirectional coupons behind the failed surface.

more scattered, they fall in the same general region. Compared to the wide variation in the laminate stress versus life data (see Fig. 7 and Ref 2), the scatter in the 0° fiber stress versus life data was much less. This indicates that the stress in the 0° fibers may be a controlling factor in laminate fatigue life.

Analytical Predictions—The stabilized stress-strain response of a $[0/90]_{2s}$ lay-up at room temperature and elevated temperature under cyclic loading is shown in Fig. 11. The curves shown were typical of lay-ups with off-axis plies. The room temperature stress-strain curve was bilinear with the unloading curve coincidental with the loading curve. The stress-strain curve at elevated temperature was nonlinear in both loading and unloading. As explained earlier, this stress-strain curve was recorded at a much slower loading rate than the majority of the fatigue cycles. Once the applied load was reduced to the minimum value, there was a net strain that was probably due to the slow loading rate. During a few of the tests, the load was held at the minimum to allow the net strain to recover. In all cases, the net strain recovered within a few minutes. The cyclic loading was then resumed. The cyclic response with the hold time is shown in Fig. 11. The most notable feature was that the initial unloading modulus was greater than the initial loading modulus.

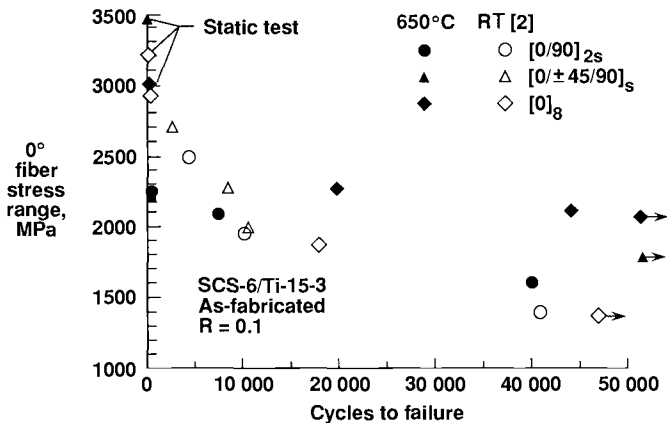


FIG. 10—Stress in the 0° fiber as a function of cycles to failure for the various layups.

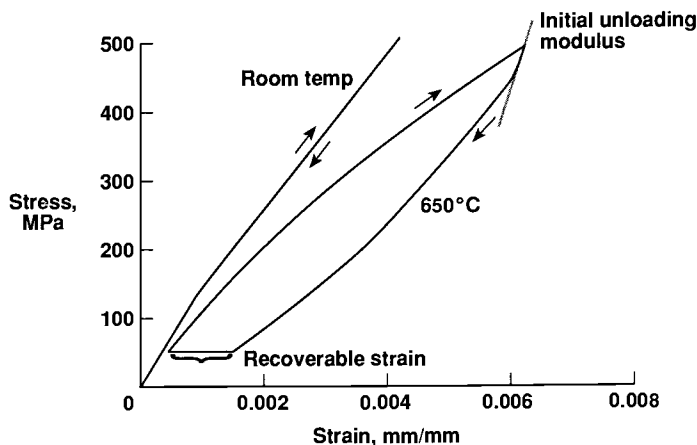


FIG. 11—Cyclic response of the $[0/90]_{2s}$ layup at room temperature and 650°C .

To understand this composite response, a coupon of plate matrix material was cycled in the same manner. Using the same strain rate and strain range, a coupon of the plate matrix material was cycled at 650°C in strain control at a rate of 2×10^{-4} mm/mm per second. After the first few cycles, the stress-strain response stabilized. The stabilized response is shown in Fig. 12. Because the response has stabilized, there are only two components of strain, a time-dependent component and elastic component. When the load is increasing, both components of strain increase. When the load is decreasing, the elastic component decreases, while the time-dependent component increases. These components sum to an initial unloading modulus that is greater than the initial loading modulus.

The stress-strain response of the $[0/90]_{2s}$ was predicted using the stress-strain response of the matrix material. The stress-strain curve for the matrix material was divided into two parts, a loading and unloading curve, and used as the matrix input for the AGLPLY program. The loading portion of the matrix stress-strain curve was idealized as perfectly elastic-plastic in the AGLPLY program. Two composite stress-strain curves were predicted, a tensile curve using the matrix loading stress-strain response and a compressive curve using the matrix unloading stress-strain response. The compressive curve was attached to the end of tensile curve to pre-

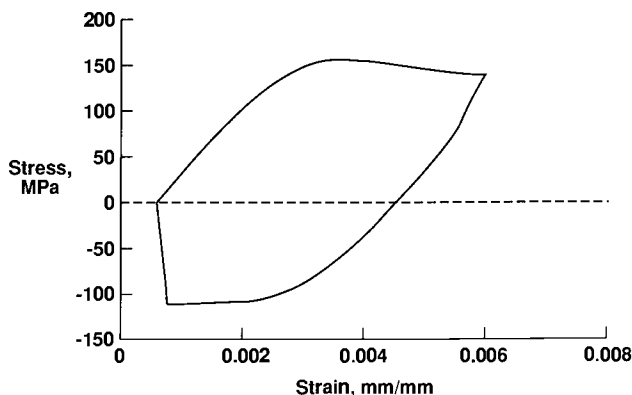


FIG. 12—Strain control cycle test of the matrix at 650°C .

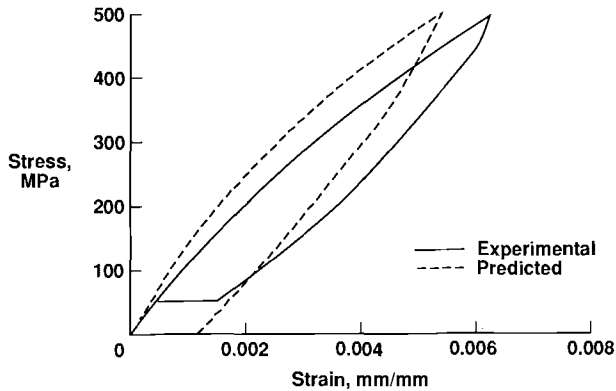


FIG. 13—Predicted and experimental cyclic stress-strain response of the $[0/90]_{2s}$ layup at 650°C .

dict the response of the $[0/90]_{2s}$ lay-up to the cyclic loading. This predicted curve is shown with the experimental curve in Fig. 13. The predicted strain levels were less than the experimental, but the trends were shown correctly. The AGLPLY program correctly predicted the initial unloading modulus to be larger than the loading modulus, and also predicted a final net strain.

Fractography

This section presents the fractography of the matrix and the composite at elevated temperature and relates it to the mechanical response for the various loading histories. The emphasis is on damage initiation and propagation.

Matrix Material

Monotonic—Tensile testing of the matrix in both load and strain control at elevated temperature resulted in ductile failure. The elongation in the load-controlled test was localized around the fracture surface. This matrix elongation caused necking, creating a ridge at the failure surface. There were dimples from the ductile failure on the ridge. A metallographic section parallel to the loading axis revealed no secondary cracks behind the fracture surface.

During the strain-controlled monotonic test at 650°C , dimples formed as a result of ductile failure. In the strain-controlled test, the coupon did not neck as in the load-controlled test.

Fatigue—In load control, the cyclic test at 650°C exhibited the same type of ductile failure surface as the monotonic test. There were dimples on the ridge formed during necking from the ductile failure. No secondary cracks behind the fracture surface were apparent in the polished section.

The fracture surface shown in Fig. 14, typical of the strain-controlled cyclic tests, exhibited a cracked region due to fatigue. The rest of the surface fractured in a ductile manner, similar to the monotonic test. Polishing perpendicular to the loading axis revealed secondary cracks for each strain range. These cracks (a typical crack is shown in Fig. 15) were, in all cases, transgranular. Fatigue crack growth occurred in the strain-controlled test because the strain control limited the amount of time-dependent deformation. In contrast, the coupon in a load-controlled test was free to creep and failed in a ductile manner.

Composite

Monotonic—All the monotonic tests of the composite were conducted in load control at 650°C . Monotonic failure of each lay-up exhibited ductile voiding around the ends of the bro-

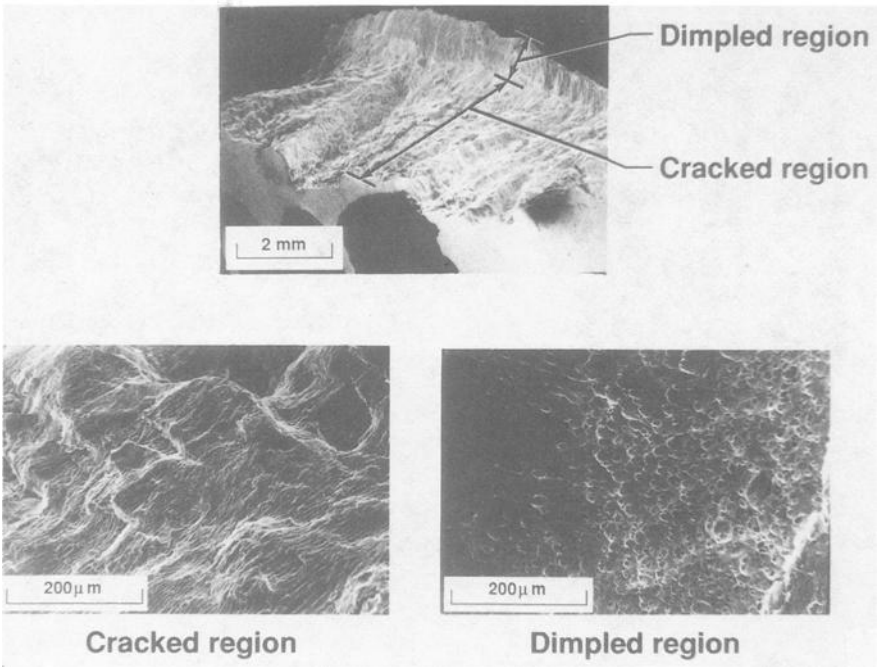


FIG. 14—Fractograph from strain-controlled fatigue test of the matrix at 650°C. Cycled 109 000 times to failure with a maximum strain of 0.0047 mm/mm.

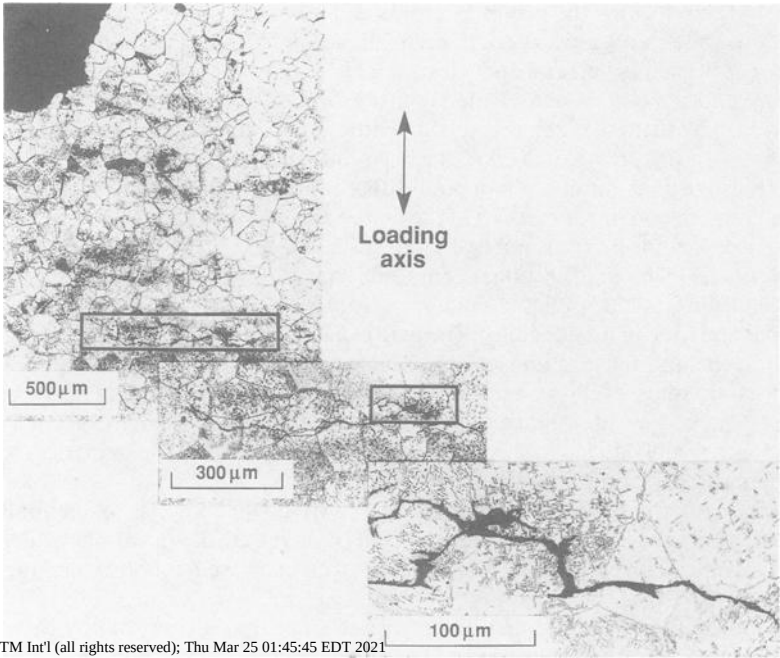


FIG. 15—Transgranular cracking of the matrix at 650°C under strain-controlled cyclic loading. The grains are visible because of the Kroll's etch. Sample was cycled 13 300 times to failure with a maximum strain of 0.0077 mm/mm.

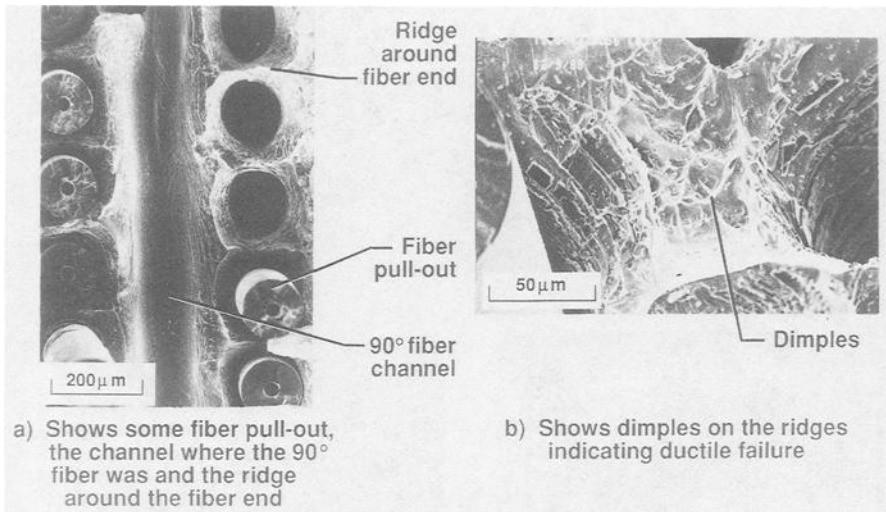


FIG. 16—Monotonic tensile failure of the $[0/90]_{2s}$ composite at 650°C .

ken 0° fibers; a typical example for $[0/90]_{2s}$ is shown in Fig. 16a. Necking of the matrix between the fibers caused the matrix to pull away from the fibers, forming ridges. Dimples were visible on these ridges indicating ductile failure of the matrix, as shown in Fig. 16b. The channel left by the 90° fiber shown in Fig. 16a was seen in all lay-ups with 90° plies. The channel was created when the interface failed and the 90° fiber was pulled away.

Fatigue—Table 1 shows the maximum stress and number of cycles to failure for each lay-up tested. First, the characteristics common to all the lay-ups are discussed. Then the specific characteristics of each lay-up and cyclic test are presented.

General characteristics—Each of the lay-ups exhibited two failure morphologies. There were flat regions with matrix fatigue cracks and little, if any, debonding along the length of the fiber as seen in the $[0]_s$ lay-up in Fig. 17. The fibers failed close to the plane of the matrix crack. There were also regions similar to monotonic ductile matrix failure, pulled-out fiber lengths, and fiber-matrix debonding. Fractions of the failure surface would be covered by each failure morphology depending on the number of cycles to failure.

Unidirectional—The unidirectional composite that was cycled at 650°C at a maximum stress of 759 MPa (183 000 cycles to failure; see Table 1) had a flat fracture surface with little fiber pullout and little matrix ductility. The matrix fatigue crack progressed across most of the cross section. Behind the fracture surface were several long matrix cracks, but few broken fibers. The strain range of this composite was below the fatigue limit of the fibers, but not of the matrix. Cracks grew long enough in the matrix eventually to break the fibers. Failure of the fibers was probably due to local stress concentrations caused by the matrix crack or local environmental attack of the fiber or both.

The fatigue test with the maximum stress of 848 MPa (44 000 cycles to failure) had a life where both the fibers and the matrix were cracking. Fatigue crack growth was evident on the fracture surface, and several matrix cracks with broken fibers were found behind the fracture surface.

The unidirectional composite that was cycled at a maximum stress of 759 MPa (20 000 cycles to failure) had many broken fibers behind the fracture surface but few matrix cracks.

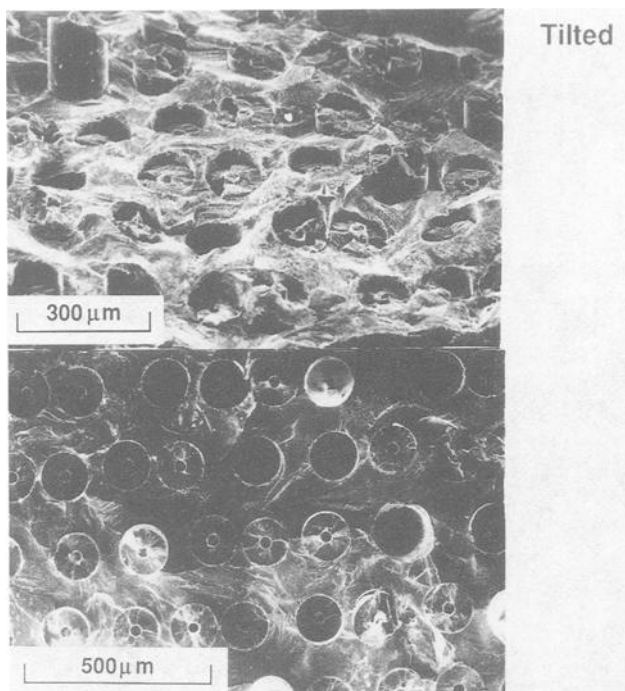


FIG. 17—Fractograph of unidirectional composite cycled 183 000 times at a maximum stress of 758 MPa. Fractograph shows the flat matrix crack and the broken fibers close to the same plane.

The fracture surface showed only two small regions with fatigue type failure. The rest of the fracture surface exhibited ductile failure with the matrix pulling away from the fibers as in the monotonic failure.

Lay-ups with off-axis plies—Damage in the lay-ups with off-axis plies initiated with interfacial failure between the matrix and fibers. As in the unidirectional composite, the amount of matrix cracking depends on the applied load range. Fracture surfaces from low maximum stress, long life tests exhibited matrix fatigue cracking. The greater the life, the more extensive the matrix cracking was.

Figure 18 shows the extent of fatigue crack propagation for a low maximum stress, high life fatigue test. This test was fatigued at a maximum stress of 337 MPa until it failed at 64 000 cycles. The cracks initiated in the 90° ply and progressed along the length of the 90° fiber. On the fracture surface shown in Fig. 18, matrix fatigue cracks were also present in the 45° plies and even in the 0° ply. The ductile regions have the same features as a monotonic failure.

Conclusions

Ti-15-3 reinforced with SCS-6 silicon carbide fibers, in five different lay-ups, was tested at 650°C to determine fatigue life, tensile strength, elastic modulus, elongation to failure, and damage initiation and progression. The elevated temperature results were compared to those previously obtained at room temperature. The monotonic stress-strain response of the laminate and a typical cyclic stress-strain curve loop were predicted using constituent properties.

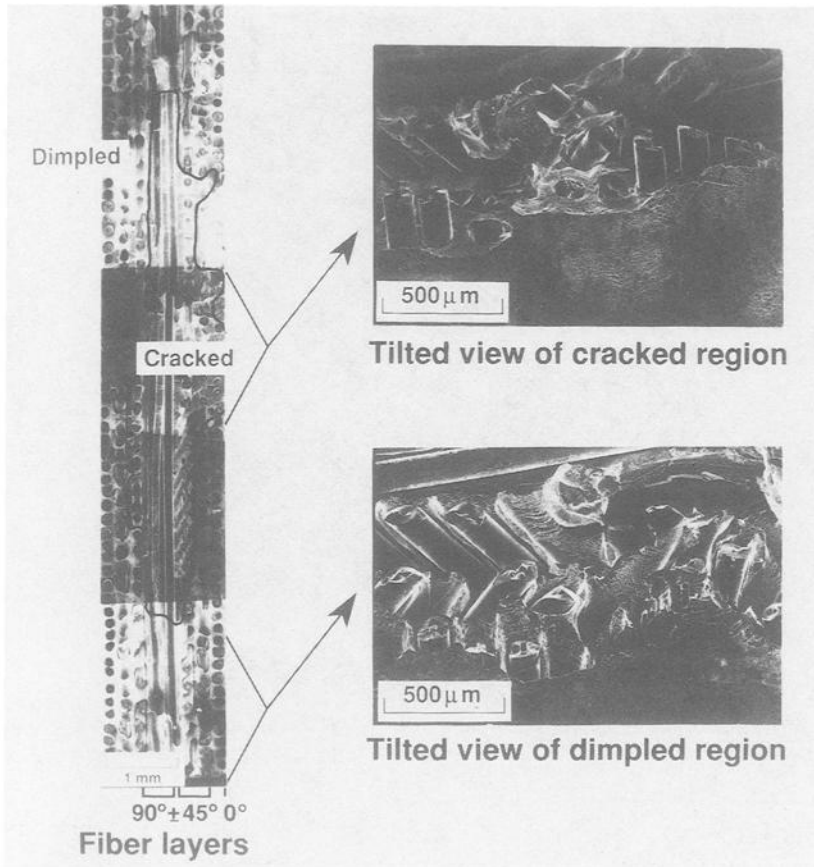


FIG. 18—Extent of fatigue cracking in a $[0/\pm 45/90]_s$ composite cycled 64 000 times at a maximum stress of 337 MPa.

Strain-controlled tests of the matrix at 650°C provided input for the predictions. Based on these tests, the following conclusions were formed:

- At elevated temperature, the fibers contributed more to the monotonic tensile strength and the fatigue endurance of the laminate than at room temperature because the matrix is relatively weak at the elevated temperature.
- The stress in the 0° fibers dominated the monotonic response of the composite. The 0° fibers contributed more to the elastic moduli than the other constituents and dominated the tensile strength and elongation to failure.
- The static strength of the in-situ fiber decreased when held for an hour and tested at 650°C, as indicated by the decrease in laminate strain to failure at temperature. This decrease in fiber strength was not observed in single-fiber tests at elevated temperature [8].
- Strain-controlled tests of the matrix material better approximated the in-situ composite response than load-controlled tests. It is important to model the matrix properly when predicting the response of the composite.

- Initial damage developed in either the fibers or the matrix. This damage was partitioned as a function of the life and applied strain range for a unidirectional lay-up. High strains and short lives resulted in only multiple fiber failure with no signs of matrix fatigue cracking. Low strains and long lives resulted in extensive matrix cracking and no fiber breakage away from the fracture surface.
- The cyclic life of the composite was governed by the cyclic stress in the 0° fibers. For each lay-up tested, the 0° fiber stress plotted against the cycles to failure at room and elevated temperature fell within a single band. More scatter was found in the 0° fiber stress versus high temperature fatigue life data than for the room temperature data.
- The initial unloading modulus was greater than the initial loading modulus. This was due to elevated-temperature time-dependent deformation of the matrix. This phenomenon was modeled using the cyclic response of the matrix material in strain control and the elastic response of the fibers.
- The matrix appeared to be too weak to cause fiber-matrix interface failure before matrix yielding when tested at 650°C. The stiffness loss that may be due to interfacial failures was obscured by the matrix yielding. The fiber-matrix interface strength may also increase after being held at the elevated temperature. Thus, the knee observed in room temperature tests was not apparent at the elevated temperature.

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High-Temperature Fatigue Behavior of a SiC/Ti-24Al-11Nb Composite

REFERENCE: Bartolotta, P. A. and Brindley, P. K., "High-Temperature Fatigue Behavior of a SiC/Ti-24Al-11Nb Composite," *Composite Materials: Testing and Design (Tenth Volume), ASTM STP 1120*, Glenn C. Grimes, Ed., American Society for Testing and Materials, Philadelphia, 1992, pp. 192-203

ABSTRACT: A series of tension-tension strain- and load-controlled fatigue tests was conducted on unidirectional SiC/Ti-24Al-11Nb (atomic %) composites at 425 and 815°C. Several regimes of damage were identified using Talreja's concept of fatigue life diagrams. Issues of test technique, test control mode, and definition of failure were also addressed.

KEYWORDS: composites, fatigue, titanium aluminide, silicon carbide, test methods

The intermetallic composite SiC/Ti-24Al-11Nb (atomic %) has generated considerable interest for aerospace applications because of its high strength-to-density ratio. Recent investigations have examined various aspects of this composite such as fiber-matrix chemical interactions [1-6], fiber-matrix bond strength [7], tensile properties [1,2,8], and thermal cycling [9,10]. However, only one study to date has examined the fatigue behavior of this composite [11]. Further study of SiC/Ti-24Al-11Nb is necessary for complete understanding of fatigue behavior and damage mechanisms. The current study examined the fatigue response of unidirectional SiC/Ti-24Al-11Nb isothermally fatigued at 425 and 815°C. The influence of test methodology (strain-versus load-control) on failure mechanisms was also a major topic of this investigation. This study is the first portion of an ongoing examination of fatigue properties and damage mechanisms in SiC/Ti-24Al-11Nb.

Experimental Procedure

The SiC/Ti-24Al-11Nb composites were fabricated by the powder cloth technique [12] followed by vacuum hot-pressing (VHP). The fiber used was the 140- μ m-diameter, double-coated, SCS-6 SiC manufactured by Textron, Inc. The matrix material was Ti-24Al-11Nb obtained as prealloyed Plasma Rotating Electrode Process (PREP) powder. The composites were produced with three plies of continuous fibers oriented in the 0° direction (parallel to the loading axis during testing) with an approximate volume fraction of 27%. Composite plates typically measured 150 mm long by 50 mm wide by 0.7 to 0.8 mm thick. Consolidated plates were etched in nitric acid to remove the Mo cladding required during VHP. Typical fiber distribution in the fully consolidated composite is shown in Fig. 1a. As shown in Fig. 1b, the composite microstructure was quite complex. The matrix was comprised of equiaxed α_2 (Ti₃Al) surrounded by disordered β . A β -depleted zone was present in the matrix adjacent to

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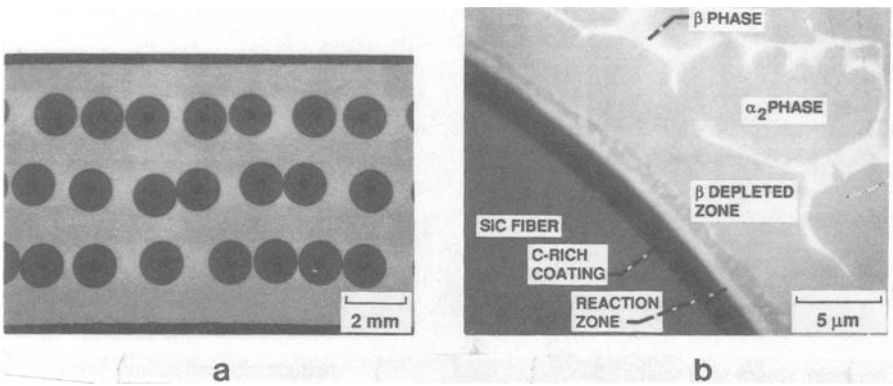


FIG. 1—As-fabricated SiC/Ti-24Al-11Nb composite showing (a) typical fiber distribution and (b) microstructure near fiber-matrix interface.

the fiber-matrix reaction zone that surrounded the C-rich coating of the fiber. A more extensive analysis of this as-fabricated composite microstructure has been previously performed by scanning transmission electron microscopy [3].

Composite specimens were wire electrodischarge machined (EDM) from the plate as shown in Fig. 2a and b. The specimen design incorporated a reduced, parallel-sided test section to facilitate close control of the temperature profile and a large fillet radius of 61 mm to minimize failures near the radius. After EDM, specimen surfaces were polished with 180-grit SiC paper to remove a 10- μ m cladding reaction with the Mo release sheets used during fabrication and to remove damage as a result of wire EDM. Tabs, as shown in Fig. 2b, were applied with epoxy. This increased the gripped-section thickness to approximately 6 mm, nearer to that of the grip

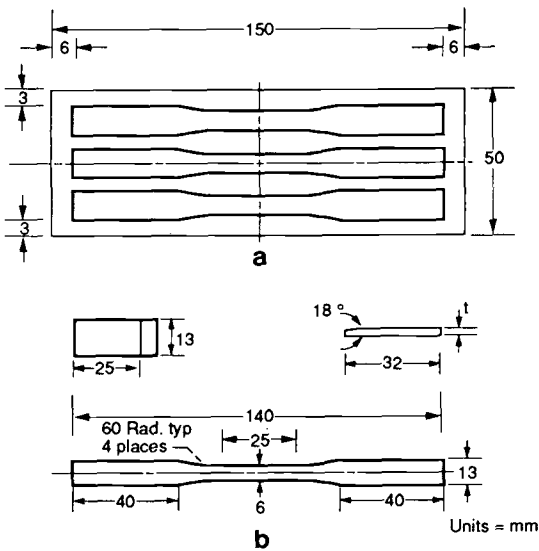


FIG. 2—Test specimen details showing (a) plate layout and (b) geometry and tab details.

opening, to help minimize possible bending stresses caused by gripping. Tabs were tapered to distribute through-thickness stresses resulting from high clamping forces. These specimen and tab designs produced fracture within the test section in 93% of the specimens.

The experiments were conducted under computer control on a closed-loop, servohydraulic system with a 222-kN maximum load capacity. A derated load capacity of 9-kN full scale was used. Water-cooled, hydraulic grips were aligned independent of the specimen to a tolerance of <3% bending during initial installation. The experimental equipment, data acquisition, and computer control system has been described previously [13]. A standard 12.7-mm gage length extensometer was mounted on the edge of the sample to avoid possible bending induced by mounting the extensometer on the width of the specimen. The specimens were heated by a two-zone direct resistance furnace specifically designed for the plate specimens [14]. This furnace maintained a temperature profile in the test section of $\pm 1\%$ of the nominal value. Beaded type K thermocouples were tied onto the test section. Specimens were maintained at temperature for 1 h before testing to allow the equipment to stabilize. Note that 1 h at the highest temperature of 815°C causes negligible growth of the fiber-matrix reaction zone [2].

To address the issue of test control method, fatigue tests were conducted in both strain-control and load-control at 425 and 815°C. Strain-controlled tests were conducted using a triangular control waveform that was limited by the required maximum strain and zero load. The strain rate was $1 \times 10^{-3} \text{ s}^{-1}$. Load-controlled tests also used a triangular control waveform in which load was limited by the maximum required load and zero load. The loading rate and maximum load limits were selected to parallel the strain rate and maximum loads achieved in the strain-controlled tests. For some of the higher load limits, a hybrid program was substituted for the load-controlled tests. The hybrid program was strain-controlled but had load limits at the maximum and minimum points of the cycle instead of strain limits. This was necessary because some of the large loads achieved in strain-controlled tests could not be achieved in load-control because the material response caused an instability in the control loop.

Results and Discussion

Fatigue Life Relationships

Fatigue life results for the SiC/Ti-24Al-11Nb composite at 425 and 815°C are presented in Fig. 3 for both strain- and load-controlled test conditions. A fatigue life diagram was created after Talreja [15] by plotting the data as maximum strain versus number of cycles to failure.

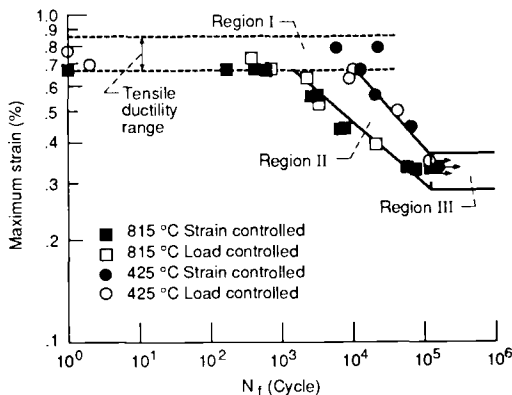


FIG. 3—Fatigue life diagram of SiC/Ti-24Al-11Nb at elevated temperatures.

Fatigue lives of the strain- and load-controlled tests were compared on a maximum strain basis because, in the absence of cracking or debonding, the strain for both the fiber and the matrix is the same during a test but the stresses in the composite constituents differ [16]. The data displayed three distinct regions, identified for this work as Regions I, II, and III. Regions I and III did not vary with test temperature while Region II was clearly temperature-dependent. These three regions were similar in shape to those of the fatigue life diagram for a polymer matrix composite at 23°C [15] as shown in Fig. 4. Note that failure of the SiC/Ti-24Al-11Nb composite specimens in Fig. 3 was defined as fracture into two pieces except for the strain-controlled tests in Region II in which failure was defined as the first simultaneous decrease in stress and stiffness of ≥ 30 MPa and 10 GPa, respectively, as will be discussed shortly.

For the SiC/Ti-24Al-11Nb composite, Region I of Fig. 3 was approximately bounded by the tensile ductility ranges for this composite, which were independent of test temperature at 425 and 815°C, at values of 0.69 to 0.85% [8]. It was previously suggested that the tensile properties of SiC/Ti-24Al-11Nb were dominated by the SiC fiber over a temperature range of 23 to 815°C [8]. This was because the composite failed at strains slightly less than or equal to the fiber strain-to-failure at all temperatures tested, while the total strain-to-failure of the matrix increased from 1.6 to 20.6% over the temperature range. Likewise, it is reasonable to suggest that fiber strain to failure also largely affected the fatigue behavior of this composite at 425 and 815°C within Region I, especially because several tests failed on load-up. Furthermore, the probabilistic failure nature of this composite in tension was attributed to variations in fiber strength (and, thus, fiber strain to failure) both between fiber lots and within a fiber lot [8]. Fiber strength variability within a given specimen would be expected to mimic the variability within a given fiber lot. Therefore, variations in fiber strengths and, thus, variations in fatigue lives, were also expected within this high strain Region I. In fact, fatigue lives for the SiC/Ti-24Al-11Nb composite ranged from $\frac{1}{4}$ cycle to approximately 700 cycles at 815°C and from $\frac{1}{4}$ cycle to approximately 20 000 cycles at 425°C in this region. This apparent dominant effect of fiber failure does not exclude other operative mechanisms.

With one exception, failure within Region I occurred by complete fracture and was defined as catastrophic or nonprogressive because the material response gave only minor indications of the impending failure. The typical maximum and minimum stress response as a function of cycles is shown in Fig. 5 for specimens at 425 and 815°C. By test design, the load-controlled tests showed a constant maximum stress until failure. A typical specimen tested in strain control at 425°C displayed a relatively constant stress response until the last few thousand cycles

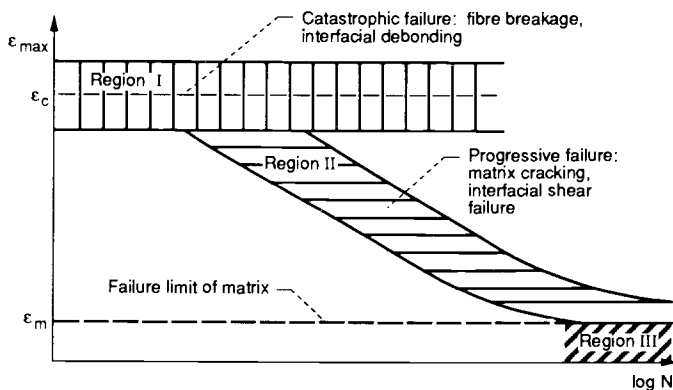


FIG. 4—Talreja's fatigue life diagram for unidirectional polymer matrix composites at room temperature.

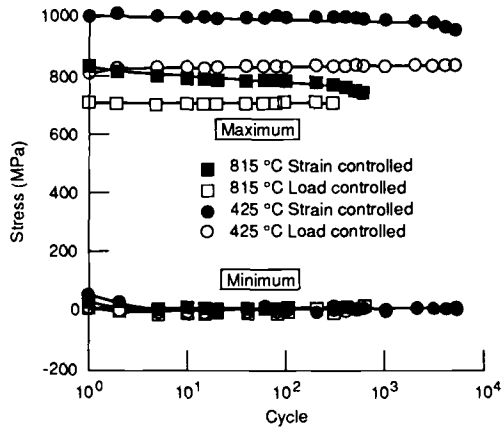


FIG. 5—Variation of maximum and minimum stress for Region I tests.

before failure in which stress slowly decreased by 40 MPa. At 815°C, typical strain-controlled specimens displayed stress decreases that totaled approximately 45 MPa in the first several cycles, as well as additional stress decreases that totaled about 55 MPa in the last several hundred cycles before failure.

Figure 6 shows maximum and minimum strain as a function of cycles for samples in Region I at 425 and 815°C. Again, by design, strain-controlled tests exhibited a constant maximum strain throughout the tests. The increases in minimum strain for the strain-controlled tests were artifacts of the control program, which did not allow the composite to go into compression and will be discussed shortly. Specimens tested in load-control exhibited ratchetting up to 0.05% strain at 425°C, which was approximately 10% of the total strain range, and up to 0.13% strain at 815°C, which was approximately 23% of the total strain range.

The elastic moduli (E) for both strain- and load-controlled specimens within Region I are shown in Fig. 7. The value of E remained constant, within experimental error, at both 425 and 815°C until the last cycles of life during which it appeared to decrease. This slight decrease in

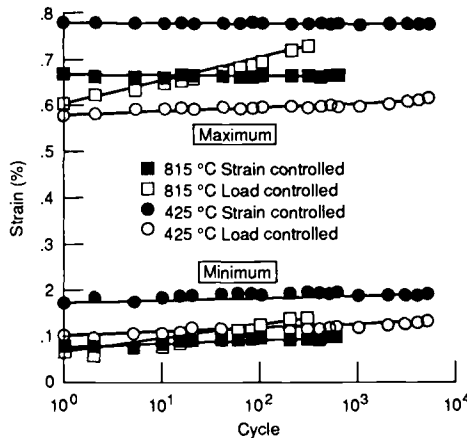


FIG. 6—Variation of maximum and minimum strain for Region I tests.

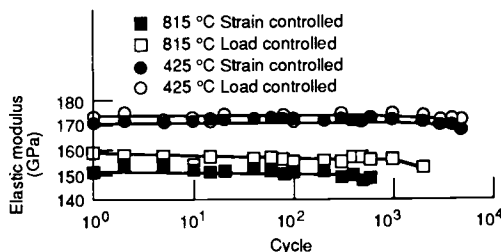


FIG. 7—Variation of elastic modulus for Region I tests.

E was most noticeable in the strain-controlled tests. An E of 168 GPa at 425°C was obtained in this study as compared with the E obtained in tension of 180 GPa [8]. E values of 151 to 157 GPa were obtained at 815°C and were slightly greater than the previous E values obtained in tension of 138 GPa [8]. One possibility for the differences in E is variation in fiber volume for samples tested in tension and those of this study; volume fraction of fibers has been determined for each individual tensile specimen [8] but is only approximately known for the samples of this study.

Fatigue behavior within Region II of Fig. 3 was defined as progressive because the mechanical response of the strain-controlled tests at both 425 and 815°C first displayed gradual decreases in the maximum stress and then abrupt, simultaneous decreases in the maximum stress (σ_{MAX}) and unloading modulus with continued cycling, as shown in Figs. 8 and 9. Furthermore, the specimens typically remained intact after these abrupt indications of damage. With these evidences of damage accumulation throughout the life of the composite, it became necessary to define failure for the strain-controlled tests in Region II as something other than final fracture, as has been suggested by Salkind [17]. Thus, failure for the strain-controlled tests in Region II of this work was defined as the cycle at which the first abrupt sign of damage was observed (for example, the first abrupt, simultaneous decrease in stress and modulus). In the lower strain portion of Region II, this initial degradation in strength and modulus was approximately 30 MPa and 10 GPa, respectively; degradation increased as the strain range was increased. However, the load-controlled specimens did not display these stress drops, as shown

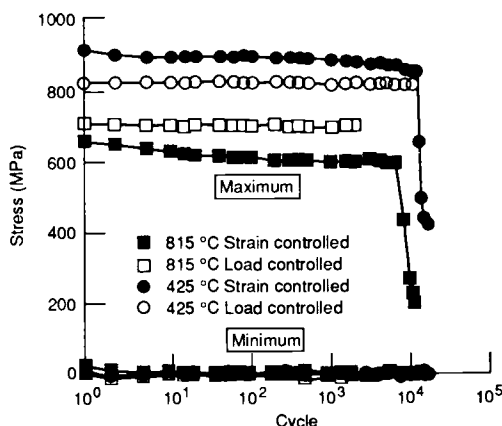


FIG. 8—Variation of maximum and minimum stress for Region II tests.

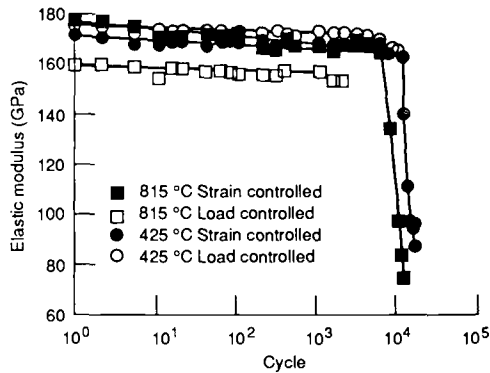


FIG. 9—Variation of elastic modulus for Region II tests.

in Fig. 8, but failed catastrophically by complete fracture of the specimen. It can be seen in Fig. 3 that the above definition of failure in strain control was a reasonable approach because, at a given temperature, failure in strain control corresponded well with complete fracture in load control. And as previously mentioned, Region II of Fig. 3 displayed temperature-dependent fatigue behavior in which specimens tested at 425°C had longer lives than those tested at 815°C at the same strain amplitude.

Figures 10 through 13 show typical stress-strain responses of the SiC/Ti-24Al-11Nb composite within Region II of Fig. 3. Comparison of the responses from the strain- and load-controlled tests under conditions with similar initial values of strain and stress at a given temperature are presented for the 425°C tests in Figs. 10 and 11 and for the 815°C tests in Figs. 12 and 13. Both test methods at 425°C resulted in approximately the same amount of inelastic strain (0.10%) after the first cycle as did both test methods at 815°C, which exhibited 0.04% inelastic strain after the first cycle. For both test methods at both temperatures, the hysteresis did not close (that is, did not become a loop) upon the first reverse cycle because neither control program allowed the composite to experience compressive stress. However, in all subsequent

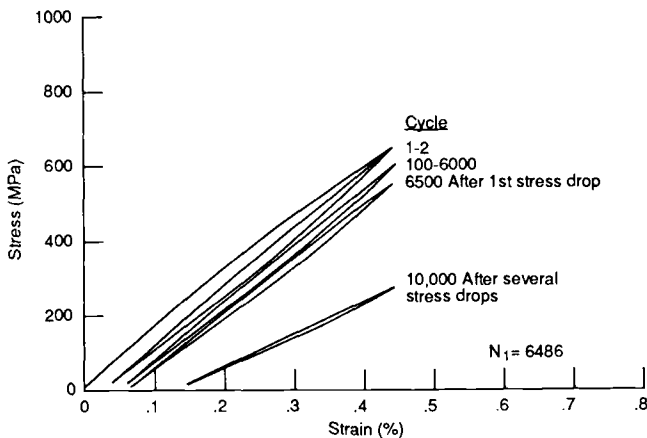


FIG. 10—Typical stress-strain response of SiC/Ti-24Al-11Nb conducted at 815°C with $\epsilon_{max} = 0.45\%$ under strain control (Region II).

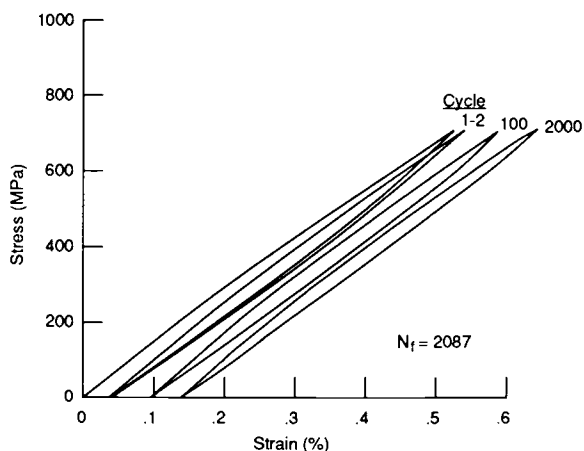


FIG. 11—Typical stress-strain response of SiC/Ti-24Al-11Nb conducted at 815°C with $\sigma_{max} = 710$ MPa under load control (Region II).

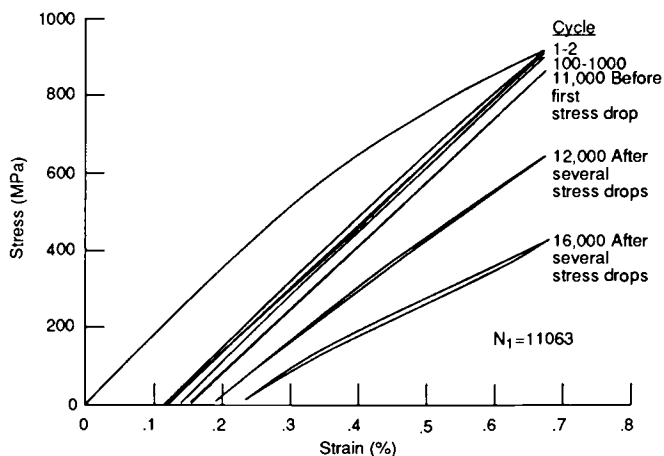


FIG. 12—Typical stress-strain response of SiC/Ti-24Al-11Nb conducted at 425°C with $\epsilon_{max} = 0.67\%$ under strain control (Region II).

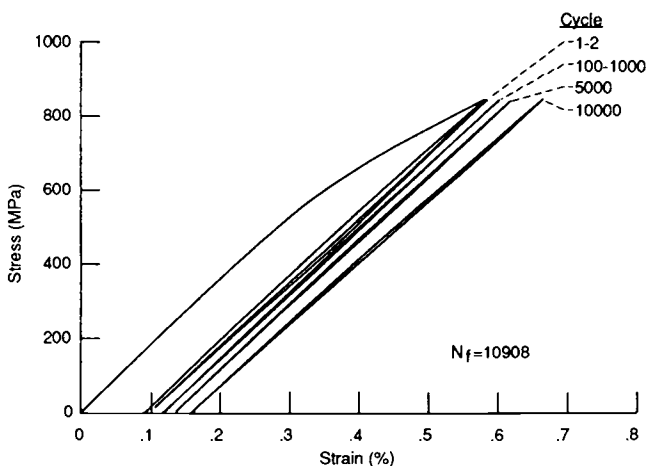


FIG. 13—Typical stress-strain response of SiC/Ti-24Al-11Nb conducted at 425°C with $\sigma_{max} = 830$ MPa under load control (Region II).

cycles, complete hysteresis loops were evident. For the 425°C strain-controlled test, the inelastic strain decreased to zero before the first abrupt stress drop and increased thereafter. The 425°C load-controlled test and the 815°C strain- and load-controlled tests all exhibited approximately constant inelastic strain (measured from hysteresis widths) from the second through final cycles at 0.005, 0.02, and 0.04%, respectively. Both the 425 and 815°C strain-controlled tests exhibited modulus degradation simultaneously with the abrupt stress decreases as previously discussed in Figs. 8 and 9. In contrast, the load-controlled tests at 425 and 815°C exhibited approximately constant modulus throughout the tests with no stiffness degradation before failure. In this discussion of fatigue response, it is important to note that the usual meaning of inelastic strain and hysteresis for monolithic materials may be different than its definition for composites. Inelastic deformation for composites may be due to one or more of events such as matrix deformation, matrix cracking, fiber/matrix debonding, or fiber fracture. It is not possible to attribute particular types of damage to inelastic behavior until metallographic work is completed. However, the abrupt decreases in strength and modulus in the strain-controlled tests can certainly be presumed to have involved at least fiber fracture to have obtained such large quantities of degradation. This reasoning relies on the facts that the fibers provide the majority of the strength and stiffness in this composite [9] and that the decreases in strength and stiffness occurred simultaneously as would be expected with fiber fracture.

Two other general observations can be made from the stress-strain responses of Figs. 10 through 13 that result from the type of test method used. In the strain-controlled tests of Figs. 10 and 12, the maximum stress of the composite decreased slowly in the cycles before the first abrupt stress drop. These stress decreases could be attributed to one or more of the following: softening, fatigue damage, or effects of the test control program. Because the test control program did not allow compressive stresses to be induced, the strain-controlled specimens actually experienced smaller strain amplitudes with each cycle, as can be seen in Figs. 10 and 12. Thus, it was possible that the test control mode was influential in continuously reducing the stress at the maximum control strain as it induced the continuously smaller strain amplitudes throughout the test. The possible contributions of material softening and decreased stress as a result of control mode could not be determined from these tests, increasing the strain amplitude after particular cycles would be necessary to characterize this behavior. In the load-controlled tests, strain ratchetting was observed up to 0.08% at 425°C, which was 18% of the total strain; strain ratchetting up to 0.12%, which was 25% of the total strain, was observed at 815°C. For both Regions I and II, strain ratchetting at 425°C was less than at 815°C, presumably because of the difference in creep rate of the matrix at the two temperatures [18].

Region III of Fig. 3 incorporates the high cycle fatigue lives of the SiC/Ti-24Al-11Nb composite with lives $\geq 10^5$ cycles. The maximum stresses and strains were so low in this regime that the specimens were typically runouts. No indications of damage, such as stress drops for the strain-controlled sample, were observed. Therefore, tests were discontinued after 10^5 cycles. The strain- and load-controlled samples at 425°C and the strain-controlled sample at 815°C appeared to establish an endurance limit at approximately 0.3% strain for this SiC/Ti-24Al-11Nb composite.

As previously mentioned, Talreja's [15] concept of the fatigue life diagram (Fig. 4) was developed for polymer matrix composites at 23°C. Three distinct regions were evident, each with a different dominant fatigue mechanism. For polymer matrix composites, Region I represented the region in which fiber fracture and interfacial debonding were the dominant mechanisms of damage. This region's width represented the scatter in the monotonic fracture strain of the composite. The scatter in this band depicted the fiber's probabilistic failure nature and was associated with low cycle fatigue lives. In Region II, matrix cracking and interfacial shear failure were the dominant mechanisms. Region III began at the fatigue limit of the matrix, and the dominant damage mechanism for polymer matrix composites in this region was matrix

microcrack nucleation. This region was associated with high cycle fatigue lives and usually resulted in test runouts.

While the fatigue lives of the SiC/Ti-24Al-11Nb composite at 425 and 815°C showed similar shapes in the three regions as compared to the polymer matrix composites, much of the damage mechanisms for the isothermally fatigued SiC/Ti-24Al-11Nb composite remain to be determined. Fatigue tests of the Ti-24Al-11Nb matrix are also necessary to estimate the contribution of matrix inelastic deformation in the composite in each of the three regions. Matrix tests are also necessary to determine whether Region III began at the fatigue limit of the matrix. However, as discussed previously, differences in fatigue response and fatigue failure imply certain mechanisms within each region. It appeared that Region I was dominated by catastrophic fiber fracture, particularly at low numbers of cycles-to-failure, although other mechanisms may also have been operative. In Region II, it appeared likely that the abrupt stress decreases observed in strain control were associated, at least in part, with local fiber fracture(s) because these stress decreases were accompanied by large modulus decreases. In Region III, fatigue life of the SiC/Ti-24Al-11Nb composite was similar to that of polymer matrix composites [15] in that only test runouts were observed in this region.

Test Control Mode

Since the development of the Manson-Coffin [19,20] relationship, strain-controlled tests have been prevalent in the fatigue testing of monolithic, polycrystalline materials. Mitchell [21] has presented several fundamental reasons for preferring strain control over load control. Two of the most important were that strain control provides a kinematic constraint that does not allow ratchetting to take place and that strain control provides a scenario most often like an actual component (that is, stress-strain gradients exist in many practical applications causing material at critical locations to be constrained to some extent).

Recent studies investigating high-temperature fatigue of metal matrix continuous fiber composites [22-24] used load control for other practical reasons. Chief among them was the fact that the materials were available only in the form of thin plates. Thus, there was the very real possibility of producing significant compressive stresses and buckling of the composite specimens as a result of geometric restraints if tested in strain control. By conducting these tests in load control, specimen buckling was alleviated but, in exchange, ratchetting was produced. In the present study, load-controlled testing also produced ratchetting of up to 19% of the total strain range at 815°C in Region II.

Another study on SiC/Ti-24Al-11Nb has used strain control without specimen buckling [12]. In the present study, cyclic tests were conducted with computer control in a manner that was strain-controlled and load-monitored. In these tests, the specimen was loaded to a maximum strain and then unloaded to zero stress. This method did not allow the specimen to experience compressive stresses and, by design, did not allow ratchetting to occur. However, as previously discussed and as shown in Figs. 10 and 12 for SiC/Ti-24Al-11Nb, the control program imposed successively smaller strain amplitudes on the composite as it controlled to zero stress on the reverse cycle. Thus, the tension-tension strain-control mode provided less and less severity as the test progressed. However, it is believed that this type of test is justified because thin plates of continuous fiber composite are not suitable for applications involving fully reversed cycling with compressive loads in the fiber direction.

In addition to the successively smaller strain ranges and the ratchetting discussed above, the type of test control mode (tension-tension strain-control versus tension-tension load-control) used in the fatigue of this SiC/Ti-24Al-11Nb composite was seen to produce other instances of different material response. The most striking differences in behavior occurred in Region II and are summarized here along with the similarities.

The primary similarity in the fatigue response of the SiC/Ti-24Al-11Nb composite in strain- and load-controlled testing occurred on the first cycle, where both types of tests produced a large quantity of inelastic strain. This inelastic strain was equal in both types of tests at equivalent strain amplitude and at a given temperature. Also, the inelastic strain of the first cycle was greater than that produced on any subsequent cycle in both types of tests. Therefore, metallographic examination of interrupted tests from either test control mode would be expected to result in the determination of equivalent mechanisms after the first cycle of either test. Furthermore, if one were primarily concerned with determining life of the composite, either strain- or load-controlled testing would generate this information, provided an appropriate definition for failure in strain-controlled tests is used.

The primary differences between test control modes obtained with SiC/Ti-24Al-11Nb occurred late in the test, at the first abrupt stress drop and beyond. Complete fracture in the load-controlled tests occurred catastrophically and coincided with the first abrupt stress drop observed in the strain-controlled tests, as shown in Fig. 8. Thus, an advantage of strain-controlled testing is that it allows easier identification of the failure mechanism by interruption and examination of a specimen after the stress drop has occurred. The other difference in test control mode involved the definition of fatigue life in strain control. Even after the first abrupt stress drop, defined as failure in this study, the composite continued to have appreciable load-carrying capacity for many thousands of cycles. If an application were most similar to the tension-tension strain-controlled tests, the SiC/Ti-24Al-11Nb composite may be considered to have a longer useful life than that indicated by the load-controlled tests.

Conclusions

- 1 A fatigue life diagram was beneficial in identifying three distinct regions of behavior in the SiC/Ti-24Al-11Nb composite. Regions I and III did not vary with test temperature, and Region II was temperature-dependent. Furthermore, distinct damage mechanisms are also expected to be defined in these regions.
- 2 SiC-Ti-24Al-11Nb fatigue failure within Region I was defined as catastrophic and appeared to be fiber-dominated, although other mechanisms may be operative. Fatigue behavior in Region II was defined as progressive because the strain-controlled tests exhibited first gradual decreases in stress and then abrupt, simultaneous decreases in strength and modulus during the life of the composite. In the high cycle fatigue regime of Region III, an apparent endurance limit at 0.3% strain was observed.
- 3 The definition of failure used for the strain-controlled tests in Region II (the first abrupt, simultaneous decrease in strength and modulus) was a suitable approach because it produced fatigue lives in close agreement with lives from load-controlled tests that were determined by complete fracture.
- 4 The advantages of using the tension-tension strain-controlled test method were that specimen buckling was prevented while still providing a strain-control type of test and stable fatigue cycling continued after large, abrupt decreases in load-carrying capacity. This allows easy interruption and inspection of fatigue damage. The disadvantage of this test method was the application of a continuously decreasing strain amplitude caused a continuous decrease in the severity of the test.
- 5 The advantage of using load-controlled testing was that failure was unambiguously defined. The disadvantages were that ratchetting was induced and stable fatigue cycling was not possible after a large reduction in the composite's load-carrying capability.
- 6 Because each type of test control mode has unique features and results in unique fatigue response, both strain- and load-controlled tests are important to the complete understanding of SiC/Ti-24Al-11Nb composite fatigue behavior.

Acknowledgment

The authors gratefully acknowledge the diligence and expertise of R. M. Shinn in the Fatigue Laboratory

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Deformation and Failure of Longitudinally Loaded Brittle-Matrix Composites

REFERENCE. Lee, J.-W. and Daniel, I. M., "Deformation and Failure of Longitudinally Loaded Brittle-Matrix Composites," *Composite Materials Testing and Design (Tenth Volume)*, ASTM STP 1120, Glenn C. Grimes, Ed., American Society for Testing and Materials, Philadelphia, 1992, pp. 204–221

ABSTRACT. A modified shear lag analysis is proposed for a unidirectional brittle-matrix composite under longitudinal tensile loading. It is assumed that the matrix, having a lower ultimate strain than the fiber, fails by transverse cracking. The analysis was conducted for a cylindrical element of matrix with a single fiber with various boundary conditions. Closed form solutions were obtained for stress distributions in the matrix and fiber as a function of applied stress and constituent properties in each case. In addition, the matrix crack density, debonded length, and reduced axial stiffness of the damaged model of two matrix cracks were obtained in closed form as a function of applied stress and constituent properties. The shear lag parameter is only a function of geometry and shear moduli of the constituents. The basic assumptions used in the analysis are that the shear stress in the fiber varies linearly and in the matrix varies according to an inverse function of a second-degree polynomial in the radial direction. The analysis can take into account residual stresses in the material.

The predicted stress-strain and stress versus matrix crack density curves were compared with experimental results for a silicon carbide/glass ceramic (SiC/CAS) composite subjected to monotonic loading. The predictions in general are in good qualitative agreement with experimental results. In particular, the saturation crack density measured is very close to the predicted value.

KEYWORDS: shear lag analysis, brittle-matrix composites, ceramic composites, fiber-matrix debonding, failure mechanisms

The reinforcement of brittle materials with high-strength fibers can result in ultimate strains substantially higher than for the monolithic brittle material. In such materials, the failure strain of the matrix is lower than that of the fibers and damage initiates with the formation of multiple matrix cracking. This is followed by partial debonding at the fiber-matrix interface and fiber fractures, which strongly depend on the properties of the interface [1]. Materials that behave in this manner include glass ceramics reinforced with SiC fibers [1–3] and cement reinforced with glass or steel fibers [4,5].

Many investigators have studied the stress transfer between fiber and matrix caused by matrix cracking and interface failure in these composites [1–10]. Aveston and Kelly [6] obtained the stress in the fiber and the interfacial shear stress by a modified shear lag analysis. The matrix crack density or crack spacing was obtained using this stress distribution as a func-

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tion of fiber diameter, elastic properties, and fiber volume fraction. Another modified shear lag model was developed by Greszczuk [7] for determining the interfacial shear property by conducting fiber pull-out tests. In this analysis, the thickness of the adhesive that bonds the matrix to the fiber was assumed to be the effective thickness of the interphase region, where the shear stress transfers the load from the matrix to the fiber. The prediction of shear modulus at the interface agrees with experimental data very well. Another shear lag model was analyzed by Hsueh [8] for a fiber/coating system. Here, the axial displacement of the coating, which can be considered as matrix, is constant in the radial direction and equal to that at the interface. The axial stress in the coating and interfacial shear stress distributions were obtained in closed form. It was found that if the ratio of the tensile strength of the coating to the interfacial shear strength is smaller than the ratio of the axial stress of the coating to the interfacial shear stress, coating segmentation occurs; otherwise, debonding of interface takes place.

A fracture mechanics approach was followed by Marshall et al. [3] for estimating the matrix stress required to initiate unstable growth of a preexisting matrix crack of finite length in a silicon carbide/lithium aluminosilicate (SiC/LAS) material system. McCartney [9] has shown the micromechanical analysis of stress transfer between fiber and matrix arising from an interface where there is perfect bonding, or where friction is governed by Coulomb's law. This semi-inverse elasticity approach allows the determination of stress distributions in the fiber and matrix and at the interface. The length of the frictional slip zone can be calculated for any value of applied strain.

The objective of this study is to develop, using a modified shear lag analysis, a simple model of stress transfer for a unidirectional brittle-matrix composite under longitudinal tensile loading. A shear lag parameter, which is usually defined in terms of an indeterminate quantity such as the effective thickness of the interphase region, is given in terms of the shear moduli and geometry of matrix and fiber. Shear stress distributions in the fiber and matrix are assumed to vary linearly and according to an inverse function of a second-degree polynomial in the radial direction, respectively. The quantities predicted are transverse crack density in the matrix, debonded length at the interface, and average axial stiffness of the damaged model as a function of applied stress and given damage state. Stress-strain, stress versus matrix crack density, and stress versus debonded length curves were obtained and correlated with the measured damage state.

Analysis

A model of a cylindrical matrix element with a single fiber is considered (Fig. 1). The element is subjected to an axial tensile load, P , as shown.

The first set of matrix cracks perpendicular to the loading direction is assumed to occur simply when the axial stress in the matrix, σ_{zm} , is equal to the tensile strength of the matrix, F_{mT} , or the axial strain of the matrix, ϵ_{zm} , is equal to the ultimate tensile strain, ϵ_{mT}^u , that is,

$$\sigma_{zm} = F_{mT} \quad \text{or} \quad \epsilon_{zm} = \epsilon_{mT}^u \quad (1)$$

This initial cracking is followed by a redistribution of stresses in the matrix and fiber. The loading and cross-sectional geometry configuration after matrix cracking is shown in Fig. 2. The equilibrium of forces of the element is obtained as

$$P = P_m + P_f + (\sigma_m V_m + \sigma_f V_f) \pi r_m^2 \quad (2)$$

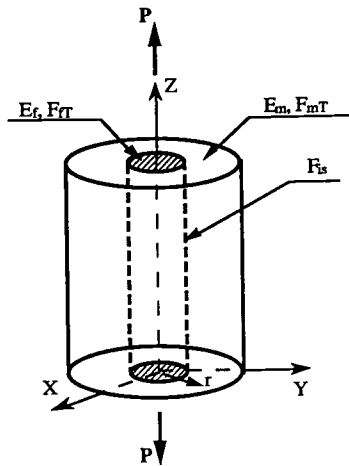


FIG. 1—Geometry and loading of cylindrical model.

where

- P = the applied force to the element,
- P_m and P_f = forces carried by the matrix and fiber,
- σ_{rm} and σ_{rf} = residual stresses in the axial direction in the matrix and fiber, respectively,
- V_m and V_f = the matrix and fiber volume ratios, and
- r_m = the radius of the cylindrical matrix element.

The basic assumptions in this analysis are that the fiber and matrix are perfectly bonded and the rate of change in the load P_m carried by the matrix is proportional to the difference in average displacements of the matrix and fiber.

$$\frac{d(P_m + \pi r_m^2 V_m \sigma_{rm})}{dz} = H(\bar{u}_m - \bar{u}_f) \tag{3}$$

where

- H = the shear lag parameter, which is a constant, and
- $\bar{u}_m$ and $\bar{u}_f$ = average axial displacements in the radial direction of the matrix and fiber, respectively.

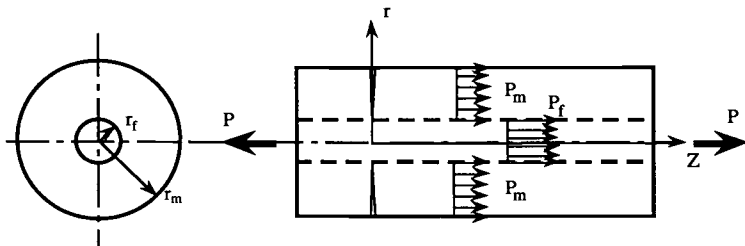


FIG. 2—Loading of cylindrical element after matrix cracking.

Therefore,

$$\frac{d^2 P'_m}{dz^2} = H \left(\frac{d\bar{u}_m}{dz} - \frac{d\bar{u}_f}{dz} \right) \quad (4)$$

where

$$P'_m = P_m + \pi r_m^2 V_m \sigma_{rm}$$

From the strain-displacement and linear stress-strain relations, we have

$$\begin{aligned} \frac{d\bar{u}_m}{dz} &= \bar{\epsilon}_m = \frac{P_m}{\pi E_m V_m r_m^2} \\ \frac{d\bar{u}_f}{dz} &= \bar{\epsilon}_f = \frac{P - P_m}{\pi E_f V_f r_m^2} \end{aligned} \quad (5)$$

where

$\bar{\epsilon}_m$ and $\bar{\epsilon}_f$ = the average axial strains in the radial direction and
 E_m and E_f = the Young's moduli of matrix and fiber, respectively

Substituting Eq 5 into Eq 4, we have

$$\frac{d^2 P'_m}{dz^2} - \alpha^2 P'_m = -\beta \quad (6)$$

where

$$\begin{aligned} \alpha^2 &= \frac{H}{\pi r_m^2} \left(\frac{1}{E_f V_f} + \frac{1}{E_m V_m} \right) \\ \beta &= \frac{H}{\pi r_m^2} \left(\frac{P}{E_f V_f} + \left(\frac{1}{E_m V_m} + \frac{1}{E_f V_f} \right) \pi r_m^2 V_m \sigma_{rm} \right) \end{aligned}$$

The general solution of Eq 6 is

$$P'_m = \frac{\beta}{\alpha^2} + R \sinh(\alpha z) + S \cosh(\alpha z) \quad (7)$$

where R and S are constants. The first term of the general solution is the force shared by the matrix when there are no matrix cracks in the element and the other terms are the disturbances due to the existence of matrix cracks.

To determine the shear lag parameter, H , we need an assumption regarding the shear stress or axial displacement variation, especially in the matrix. The outcome of the analysis is critically dependent on this type of assumption. Some investigators have assumed a uniform shear stress along the radial direction, which is valid for thin coatings [8]. Others have assumed an inverse linear distribution, which leads to unrealistic results [4]. Near (or at) the matrix crack, the axial displacement distribution in the radial direction is not linear. Guided by the expected shape of the crack [4], experimental evidence [1], and results of a finite element analysis, we assumed a linear variation of shear stress in the fiber and an inverse quadratic variation of the shear stress in the matrix in the radial direction,

$$\begin{aligned}\tau_{rzf} &= G_f(2C_1r + C_2) \\ \tau_{rz m} &= G_m \left(-\frac{C_4}{r^2} + C_5 \right)\end{aligned}\quad (8)$$

where

C_i = undetermined parameters that are functions of z ,
 G_m and G_f = shear moduli, and
 $\tau_{rz m}$ and $\tau_{rz f}$ = shear stresses in the matrix and fiber, respectively

It follows that the axial displacements in the fiber and matrix are

$$\begin{aligned}u_f &= C_1r^2 + C_2r + C_3 \\ u_m &= C_4\frac{1}{r} + C_5r + C_6\end{aligned}\quad (9)$$

where

C_3 and C_6 = undetermined parameters that are functions of z and
 u_m and u_f = axial displacements in the matrix and fiber, respectively

The boundary conditions for shear stress and axial displacement of the element with the matrix cracks subjected to the longitudinal tensile loading are (see Fig 2)

$$\tau_{rz} = 0 \quad \text{at} \quad r = 0 \quad \text{and} \quad r = r_m \quad (10a)$$

$$\tau_{rzf} = \tau_{rz m} \quad \text{at} \quad r = r_f \quad (10b)$$

$$u_f = u_m \quad \text{at} \quad r = r_f \quad (10c)$$

Using boundary conditions (Eqs 10a,b,c) and introducing the interfacial shear stress, τ_i , which will be determined later, we can get a relation as follows

$$\bar{u}_m - \bar{u}_f = A\tau_i \quad (11)$$

where

$$A = \frac{r_f}{4G_f} + \frac{1}{G_m} \left[\frac{2r_f^2(r_f - r_m)}{3(r_m^2 - r_f^2)^2} (4r_m^2 + r_m r_f + r_f^2) + \frac{r_f(r_m^2 + r_f^2)}{r_m^2 - r_f^2} \right] \quad (12)$$

From equilibrium of axial forces in the matrix of the element (Fig 3), we obtain the relation

$$\tau_i = \frac{1}{2\pi r_f} \frac{dP'_m}{dz} \quad (13)$$

Combining Eqs 3, 11, and 13, we find the shear lag parameter as a function of shear moduli of constituents and geometry of the element

$$H = \frac{2\pi r_f}{A} \quad (14)$$

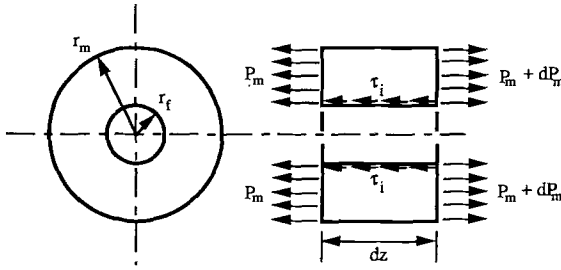


FIG. 3—Stresses in the matrix of cylindrical element.

From Eqs 2, 7, and 13, general stress distributions in an element between two matrix cracks can be expressed as follows:

$$\begin{aligned}\sigma_{zf} &= \left(\frac{E_f}{E_l} \sigma_a + \sigma_{zf} \right) + K_3 \sinh(\alpha z) + K_4 \cosh(\alpha z) \\ \sigma_{zm} &= \left(\frac{E_m}{E_l} \sigma_a + \sigma_{zm} \right) + K_1 \sinh(\alpha z) + K_2 \cosh(\alpha z) \\ \tau_i &= K_5 \sinh(\alpha z) + K_6 \cosh(\alpha z)\end{aligned}\quad (15)$$

where

$$\begin{aligned}E_l &= E_f V_f + E_m V_m = \text{the initial longitudinal modulus,} \\ \sigma_{zf} \text{ and } \sigma_{zm} &= \text{the axial stresses in the fiber and matrix, respectively, and} \\ K_1, \dots, K_6 &= \text{constants to be determined from appropriate boundary conditions.}\end{aligned}$$

Several cases are discussed below for various boundary conditions. In all cases, perfect bond between fiber and matrix is assumed.

Case 1: Embedded Fiber of Finite Length in Infinite Matrix Medium

A fiber of radius r_f and length l is embedded in an infinite matrix material (Fig. 4). An axial tensile force, P , is applied at the end of the fiber. The matrix volume ratio, V_m , is taken as 1 because the radius of the matrix is infinite. Neglecting residual stresses in the matrix and fiber, the boundary conditions for this case are:

$$P'_m = 0 \quad \text{at} \quad z = 0 \quad (16a)$$

$$P'_m = P \quad \text{at} \quad z = l \quad (16b)$$

Because $V_m \cong 1$ and $r_m \rightarrow \infty$, Eq 5 becomes

$$\bar{\epsilon}_m \approx 0 \quad \text{and} \quad \bar{\epsilon}_f = \frac{P - P_m}{\pi E_f r_f^2} \quad (17)$$

The shear lag parameter, H , is determined as follows:

$$H = \frac{8\pi G_f G_m}{4G_f + G_m} \quad (18)$$

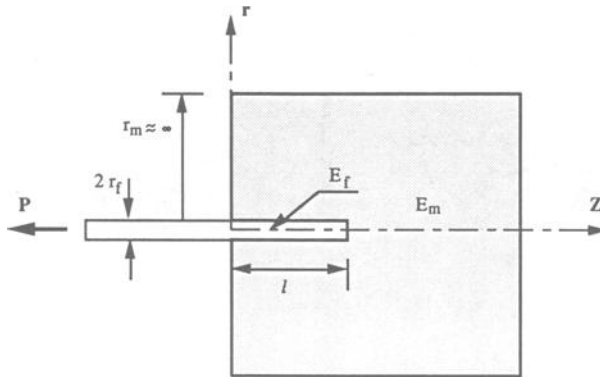


FIG. 4—Embedded fiber of finite length in infinite matrix medium.

Also, the coefficients of Eq 6 become

$$\alpha^2 = \frac{H}{\pi E_f r_f^2} \quad \text{and} \quad \beta = \frac{HP}{\pi E_f r_f^2} \tag{19}$$

Using boundary conditions (Eqs 16a,b), solution (Eq 7) takes the form:

$$P'_m = P[1 + \coth(\alpha l) \sinh(\alpha z) - \cosh(\alpha z)] \tag{20}$$

The distributions of the axial stress in the fiber and the interfacial shear stress are obtained from Eqs 2, 13, and 20 as follows:

$$\sigma_{zf} = \sigma_a [\cosh(\alpha z) - \coth(\alpha l) \sinh(\alpha z)] \tag{21}$$

$$\tau_i = \frac{P\alpha}{2\pi r_f} [\coth(\alpha l) \cosh(\alpha z) - \sinh(\alpha z)] \tag{22}$$

where σ_a is the applied stress on the fiber.

Case 2: Embedded Fiber of Infinite Length in Semi-Infinite Matrix Medium

A continuous fiber of radius r_f is embedded in a semi-infinite matrix of finite radius r_m (Fig. 5). As in Case 1, the axial tensile force, P , is applied at the end of the fiber. The boundary conditions for this case are:

$$P'_m = 0 \quad \text{at} \quad z = 0 \tag{23a}$$

$$P'_m = P \frac{E_m V_m}{E_f} + \pi r_m^2 V_m \sigma_{rm} \quad \text{at} \quad z = \infty \tag{23b}$$

Using the boundary conditions (Eqs 23a, b), solution (Eq 7) is expressed as follows:

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$$P'_m = \left(P \frac{E_m V_m}{E_f} + \pi r_m^2 V_m \sigma_{rm} \right) (1 - e^{-\alpha z}) \tag{24}$$

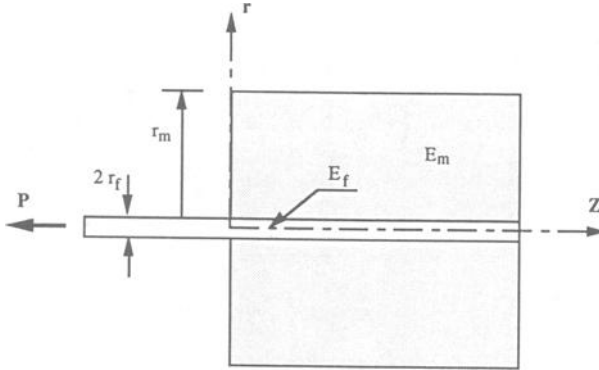


FIG. 5—Embedded fiber of infinite length in a semi-infinite matrix medium.

The axial stress distributions in the matrix and fiber and the interfacial shear stress distribution are obtained from Eqs 2, 13, and 24 as follows:

$$\sigma_{zm} = \left(\sigma_a \frac{E_m}{E_1} + \sigma_{rm} \right) (1 - e^{-\alpha z}) \quad (25)$$

$$\sigma_{zf} = \frac{\sigma_a}{V_f} \left(\frac{E_f V_f + E_m V_m e^{-\alpha z}}{E_1} \right) + \sigma_{\eta} (1 - e^{-\alpha z}) \quad (26)$$

$$\tau_i = \frac{\alpha e^{-\alpha z}}{2\pi r_f} \left(\sigma_a \frac{E_m}{E_1} + \sigma_{rm} \right) \quad (27)$$

Case 3: Fiber in Cylindrical Matrix with Two Transverse Matrix Cracks

As shown in Fig. 6, a cylindrical element of matrix with a single fiber between two matrix cracks is subjected to the axial force P . When the matrix cracks on planes $z = 0$ and l , the following boundary conditions must be satisfied:

$$P'_m = 0 \quad \text{at} \quad z = 0 \quad \text{and} \quad l \quad (28)$$

where l is the matrix crack spacing.

Using boundary condition (Eq 28), solution (Eq 7) becomes

$$P'_m = \left(\frac{E_m V_m}{E_1} P + \pi r_m^2 V_m \sigma_{rm} \right) \left[1 - \frac{\cosh \left(\frac{\alpha l}{2} - \alpha z \right)}{\cosh \left(\frac{\alpha l}{2} \right)} \right] \quad (29)$$

from which we obtain

$$\sigma_{zm} = \left(\frac{E_m \sigma_a}{E_1} + \sigma_{rm} \right) \left[1 - \frac{\cosh \left(\frac{\alpha l}{2} - \alpha z \right)}{\cosh \left(\frac{\alpha l}{2} \right)} \right] \quad (30)$$

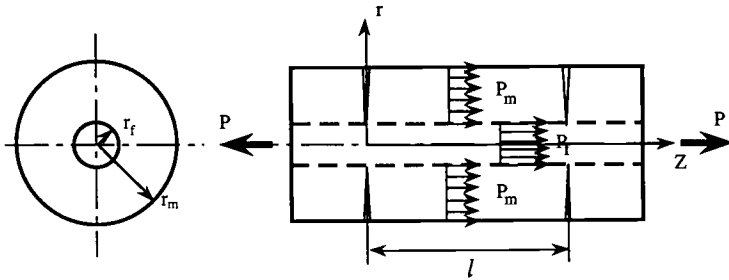


FIG. 6—Cylindrical model of a fiber with a pair of matrix cracks.

The axial stress in the fiber, σ_{zf} is obtained using equilibrium condition (Eq 2) and Eq 29 as follows:

$$\sigma_{zf} = \frac{E_f}{E_1} \left[1 + \frac{E_m V_m}{E_f V_f} \frac{\cosh\left(\frac{\alpha l}{2} - \alpha z\right)}{\cosh\left(\frac{\alpha l}{2}\right)} \right] \sigma_a + \left[1 - \frac{\cosh\left(\frac{\alpha l}{2} - \alpha z\right)}{\cosh\left(\frac{\alpha l}{2}\right)} \right] \sigma_{mf} \tag{31}$$

The interfacial shear stress distribution at a given applied stress σ_a and crack spacing l is obtained from Eqs 13 and 29 as follows:

$$\tau_i(z) = \frac{1}{2\pi r_f} \frac{dP'_m}{dz} = \frac{\alpha r_m^2 V_m}{2r_f} \left(\frac{E_m \sigma_a}{E_1} + \sigma_{mf} \right) \frac{\sinh\left(\frac{\alpha l}{2} - \alpha z\right)}{\cosh\left(\frac{\alpha l}{2}\right)} \tag{32}$$

One of the assumptions used in this analysis so far is the existence of perfect bond between the fiber and matrix. However, when the interfacial shear strength, F_{is} , is reached by the interfacial shear stress, τ_i , debonding occurs. The extent of debonding, d , is calculated as follows (Fig. 7):

Neglecting interfacial friction in the debonded area, the redistributed stresses are as follows:

In the debonded area ($0 \leq z \leq d$ and $l - d \leq z \leq l$)

$$\sigma_{zf} = \frac{\sigma_a}{V_f} \tag{33}$$
$$\sigma_{zm} = 0$$

In the bonded area ($d \leq z \leq l - d$)

$$\sigma_{zf} = \frac{E_f}{E_1} \left\{ 1 + \frac{E_m V_m}{E_f V_f} \frac{\cosh \left[\frac{\alpha l'}{2} - \alpha(z - d) \right]}{\cosh \left(\frac{\alpha l'}{2} \right)} \right\} \sigma_a \quad (34a)$$

$$+ \left\{ 1 - \frac{\cosh \left[\frac{\alpha l'}{2} - \alpha(z - d) \right]}{\cosh \left(\frac{\alpha l'}{2} \right)} \right\} \sigma_{rf}$$

$$\sigma_{zm} = \left(\frac{E_m \sigma_a}{E_1} + \sigma_{rm} \right) \left\{ 1 - \frac{\cosh \left[\frac{\alpha l'}{2} - \alpha(z - d) \right]}{\cosh \left(\frac{\alpha l'}{2} \right)} \right\} \quad (34b)$$

$$\tau_i = \frac{\alpha r_f^2 V_m}{2 r_f} \left(\frac{E_m \sigma_a}{E_1} + \sigma_{rm} \right) \frac{\sinh \left[\frac{\alpha l'}{2} - \alpha(z - d) \right]}{\cosh \left(\frac{\alpha l'}{2} \right)} \quad (34c)$$

where $l' = l - 2d$

The debonded length, d , is calculated by assuming that the shear stress at the end of the interfacial disbond is equal to the interfacial shear strength. Thus, from Eq. 34c we obtain

$$d = \frac{1}{2} \left(l - \frac{1}{\alpha} \log \frac{1 + \xi}{1 - \xi} \right) \quad (35)$$

where

$$\xi = \frac{2 F_{is}}{\alpha r_f} \frac{V_f}{V_m} \frac{E_1}{E_m \sigma_a + E_1 \sigma_{rm}}$$

The average axial stresses in the matrix and fiber with debonded length, d , from one end of the element are as follows (see Fig. 7)

$$\begin{aligned} \bar{\sigma}_{zm} &= \frac{1}{l} \int_0^l \sigma_{zm} dz \\ &= \frac{1}{l} \int_d^{l-d} \sigma_{zm} dz \\ &= \left(\frac{E_m \sigma_a}{E_1} + \sigma_{rm} \right) \left[\frac{l'}{l} - \left(\frac{2}{\alpha l} \right) \tanh \left(\frac{\alpha l'}{2} \right) \right] \end{aligned} \quad (36a)$$

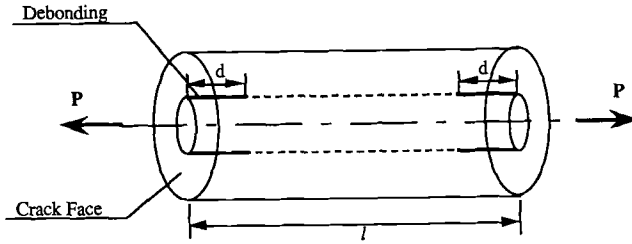


FIG. 7—Geometry and loading of a cylindrical model with partial debonding at the interface.

$$\begin{aligned}\bar{\sigma}_{zf} &= \frac{1}{l} \int_0^l \sigma_{zf} dz \\ &= \frac{1}{l} \left(\int_0^d \sigma_{zf} dz + \int_d^{l-d} \sigma_{zf} dz + \int_{l-d}^l \sigma_{zf} dz \right) \\ &= \left\{ \frac{2d}{V_f l} + \frac{E_f}{E_f V_f + E_m V_m} \left[\frac{l'}{l} + \frac{E_m V_m}{E_f V_f} \left(\frac{2}{\alpha l} \right) \tanh \left(\frac{\alpha l'}{2} \right) \right] \right\} \sigma_a \\ &\quad + \left[\frac{l'}{l} - \left(\frac{2}{\alpha l} \right) \tanh \left(\frac{\alpha l'}{2} \right) \right] \sigma_f\end{aligned}\quad (36b)$$

The average axial strain in the fiber, $\bar{\epsilon}_{zf}$, is obtained by dividing the average fiber stress above by the fiber modulus and is assumed to be equal to the average axial strain in the element, ϵ_a .

$$\begin{aligned}\epsilon_a \cong \bar{\epsilon}_{zf} &= \frac{\sigma_a}{E_f} \left\{ \frac{2d}{V_f l} + \frac{E_f}{E_f V_f + E_m V_m} \left[\frac{l'}{l} + \frac{E_m V_m}{E_f V_f} \left(\frac{2}{\alpha l} \right) \tanh \left(\frac{\alpha l'}{2} \right) \right] \right\} \\ &\quad + \frac{\sigma_f}{E_f} \left[\frac{l'}{l} - \left(\frac{2}{\alpha l} \right) \tanh \left(\frac{\alpha l'}{2} \right) \right]\end{aligned}\quad (37)$$

Thus, the reduced axial stiffness E'_1 , for a crack spacing l , and debonded length d from the crack, is obtained as follows:

$$E'_1 = \frac{E_f}{\frac{2d}{V_f l} + \frac{E_f}{E_f V_f + E_m V_m} \left[\frac{l'}{l} + \frac{E_m V_m}{E_f V_f} \left(\frac{2}{\alpha l} \right) \tanh \left(\frac{\alpha l'}{2} \right) \right] + \frac{\sigma_f}{\sigma_a} \left[\frac{l'}{l} - \left(\frac{2}{\alpha l} \right) \tanh \left(\frac{\alpha l'}{2} \right) \right]}\quad (38)$$

where σ_a is the applied stress that produces the average crack spacing l under monotonic loading, obtained from Eq 30 by replacing σ_{zm} and z by F_{mT} and $l/2$.

Stress-Strain and Stress Versus Crack Density Relations

The deformation and progressive damage in the form of stress-strain and stress versus crack density relations can be calculated from the basic material properties of the element following the steps of the flow chart of Fig. 8.

1. The stress-strain behavior is linear up to an applied stress level where the matrix stress or strain reach the ultimate values of matrix tensile strength or ultimate tensile strain (Eq 1). This stress level is given by

$$\sigma_u = \frac{E_1}{E_m} F_{mT}\quad (39)$$

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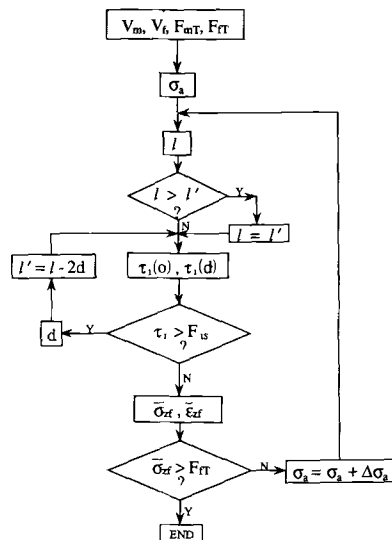


FIG 8—Flow chart for the prediction of stress-strain and stress versus crack density

- 2 Above this applied stress level, matrix cracking occurs, reaching a crack spacing l obtained from Eq 30 by setting $z = l/2$ and $\sigma_{zm} = F_{mT}$.

$$l = \frac{2}{\alpha} \cosh^{-1} \left[\frac{E_m \sigma_a + E_1 \sigma_{rm}}{E_m \sigma_a + E_1 (\sigma_{rm} - F_{mT})} \right] \quad (40)$$

- 3 The maximum interfacial shear stress, τ_i , at the crack location is calculated from Eq 32 by setting $z = 0$.

$$\tau_i(0) = \frac{\alpha r_m^2 V_m}{2r_f} \left(\frac{E_m \sigma_a}{E_1} + \sigma_{rm} \right) \tanh \frac{\alpha l}{2} \quad (41)$$

- 4 The above shear stress is compared with the interfacial shear strength F_{is}
- If $\tau_i(0) > F_{is}$, calculate debonded length, d , by Eq 35, determine $l' = l - 2d$, and calculate τ_i by Eq 34c
 - If $\tau_i(0) \leq F_{is}$, go to next step
5. The average axial stress and strain in the fiber, $\bar{\sigma}_{zf}$ and $\bar{\epsilon}_{zf}$, are calculated by Eqs 36b and 37 and $\bar{\sigma}_{zf}$ is compared with fiber tensile strength, F_{fT}
- If $\bar{\sigma}_{zf} > F_{fT}$, failure occurs
 - If $\bar{\sigma}_{zf} < F_{fT}$, go to next step
6. Applied stress is increased by $\Delta\sigma_a$
7. New crack spacing l is determined and compared with $l' = l - 2d$ from Step 4
- If $l < l'$, proceed from Step 4 until failure
 - If $l \geq l'$, no additional matrix cracking occurs at this loading step, keep $l = l'$ and repeat Step 4

Discussion

The proposed application of shear lag theory with a linear variation of shear stress in the fiber and an inverse second-order polynomial variation of shear stress in the matrix in the radial direction allows the determination of stress distributions in closed form for three different cases

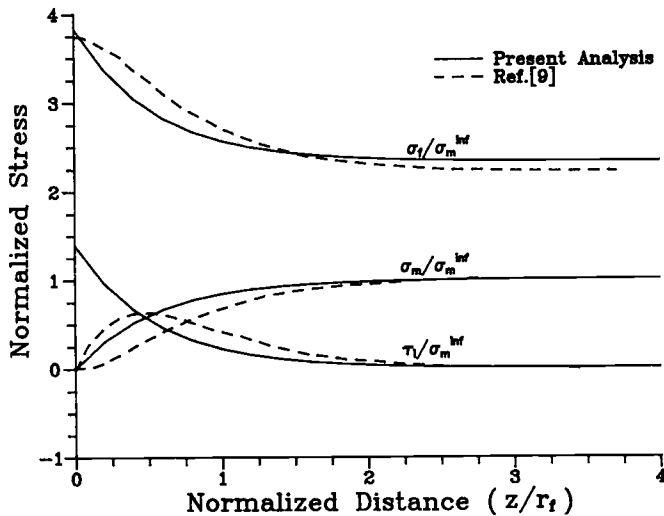


FIG. 9—Stress distributions at interface of embedded fiber of finite length in semi-infinite matrix.

For the case of an embedded fiber of finite length in an infinite matrix, the interfacial shear stress distribution of Eq 20 has the same form as that given by Greszczuk in Ref 7. In Ref 7, the parameter α depends on the ratio of shear modulus of the interface to the effective thickness of the interface, which can be obtained indirectly and is not a physical property of the material. However, in the present analysis, the parameter α is obtained simply from the elastic properties of the fiber and matrix and the radius of the fiber (see Eq 19).

For the case of an embedded fiber of finite length in a semi-infinite matrix material, the resulting stress distributions, Eqs 24 through 27, are compared with those obtained by a semi-inverse elasticity approach in Ref 9 (see Fig. 9). The elastic material properties, thermal properties, and geometry used in the prediction are as follows:

$E_f = 220 \text{ GPa (31.9 Msi)}$
 $E_m = 70 \text{ GPa (10.2 Msi)}$
 $\alpha_f = 5.5 \times 10^{-6}/^{\circ}\text{C (}3.1 \times 10^{-6}/^{\circ}\text{F)}$
 $\alpha_m = 10.0 \times 10^{-6}/^{\circ}\text{C (}5.6 \times 10^{-6}/^{\circ}\text{F)}$
 $V_f = 0.4$
 $\nu_f = 0.1$
 $\nu_m = 0.2$
 $\Delta T = -550^{\circ}\text{C}$

where

α_f and α_m = the thermal expansion coefficients of the fiber and matrix, respectively, and
 ΔT = the temperature difference between the stress-free temperature and current temperature.

From the rule of mixtures, the longitudinal thermal expansion coefficient of the composite, α_c , and the residual thermal stresses in the fiber and the matrix are

$\alpha_c = 7.0 \times 10^{-6}/^{\circ}\text{C (}3.9 \times 10^{-6}/^{\circ}\text{F)}$
 $\sigma_f = -175 \text{ MPa (-25.3 ksi)}$
 $\sigma_m = 117 \text{ MPa (16.9 ksi)}$

The applied strain on the cylindrical element is 0.8%, which is equivalent to an applied stress of 1.04 GPa (151 ksi). Using the given conditions, the stress distributions in the fiber and matrix and at the interface normalized by the undisturbed axial stress in the matrix are plotted against the axial distance normalized by the fiber radius. As can be seen in Fig. 9, the predicted axial stress distributions in the fiber and matrix and the interfacial shear stress agree with those from Ref 9 in the region of $z \geq r_f$. However, the interfacial shear stress from the present analysis increases as the distance z from the crack decreases, while that from Ref 9 has a maximum value at $z \approx 0.5r_f$ and decreases to zero at the crack face. Because the axial stresses in the fiber and matrix are related to the interfacial shear stress as seen from Eqs 2 and 13, this difference in the interfacial shear stress distribution causes the deviation between both predictions of the axial stress distributions in the fiber and matrix in the region of $0 \leq z \leq 0.5r_f$.

For the case of a fiber with a set of two matrix cracks subjected to axial loading (Case 3), a comparison was made of the axial stress in the matrix and interfacial stress distribution with the results of another shear lag approach (see Ref 8). In Ref 8, Hsueh assumed the displacement of the matrix to be constant in the radial direction and equal to that at the interface. The configuration and material properties used for this comparison are

$$\begin{aligned} \frac{E_m}{E_f} &= 0.8 & \frac{r_m}{r_f} &= 1.2 \\ \nu_f = \nu_m &= 0.3 & \frac{l}{r_f} &= 10 \end{aligned}$$

As can be seen in Fig. 10, there is no difference between the two predictions except for the maximum normalized value of the interfacial shear stress. The normalized interfacial shear stress predicted by the present analysis is 0.39; Hsueh predicted a value of 0.45.

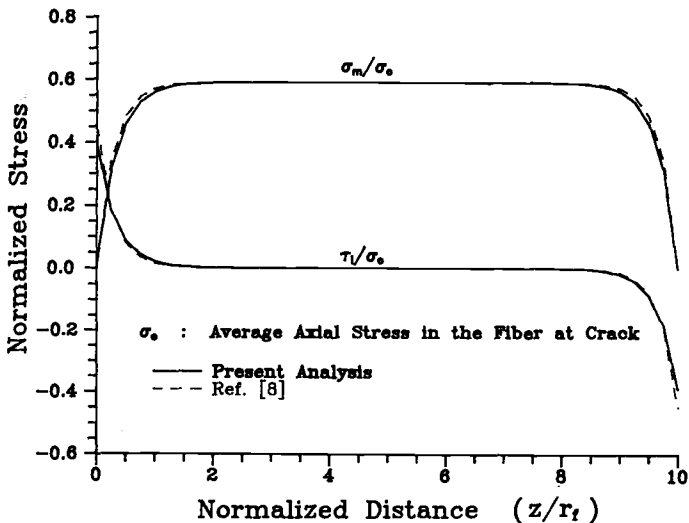


FIG. 10—Distribution of axial stress in matrix (or coating) and interfacial shear stress for a fiber embedded in matrix with two transverse cracks.

TABLE 1—Constituent material properties.

Property	CAS Matrix ^a	SiC Fiber
Maximum use temperature, °C (°F)	1350 (2460)	1300 (2370)
Fiber diameter, μm	...	16
Density, g/cm ³	2.8	2.6
Coefficient of thermal expansion, 10 ⁻⁶ /°C (10 ⁻⁶ /°F)	5.0 (2.8)	3.1 (1.7)
Elastic modulus, GPa (Msi)	98 (14.2)	179 (26)
Tensile strength, MPa (ksi)	124 (18) Flexural	2060 (300)

^aD. C. Larsen, manager, Materials Research, Corning Glass Works, Corning, NY, 1988, private communication.

The proposed theory is applied to a SiC/CAS composite, calcium aluminosilicate glass ceramic reinforced with silicon carbide fibers, to check the stress distributions for various crack spacings. The elastic and thermal properties of the SiC/CAS material system are given in Table 1.

Figure 11 shows a typical stress-strain curve to failure obtained from a tensile test of this material and compared with the predicted stress-strain curve obtained from the present analysis. In the prediction, the matrix tensile strength used was 159 MPa (23 ksi), which is higher than that given in Table 1. This is justified by the fact that the matrix tensile strength in the composite is usually higher than that of the pure matrix material [10]. The interfacial shear strength of this material system was assumed to be 221 MPa (32 ksi) to match the prediction with experimental results, because the interfacial properties are not known yet.

From Fig. 11 the predicted stress-strain curve agrees qualitatively with the experimental one. The experimental stress-strain curve shows several characteristic features that are related to the failure mechanisms and failure process. The initial Region AB corresponds to the linear elastic behavior of the material before any significant failure. The measured modulus is $E_1 = 125$ GPa (18.1 Msi). The prediction by the rule of mixtures is 130 GPa (18.8 Msi). At an applied stress of approximately 207 MPa (30 ksi) (Point B), there is a marked departure from

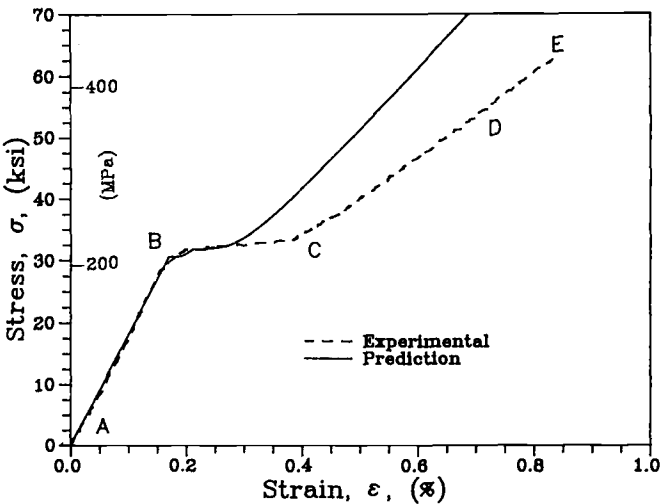


FIG. 11. Stress-strain curves for SiC/CAS [0_s] unidirectional composite.

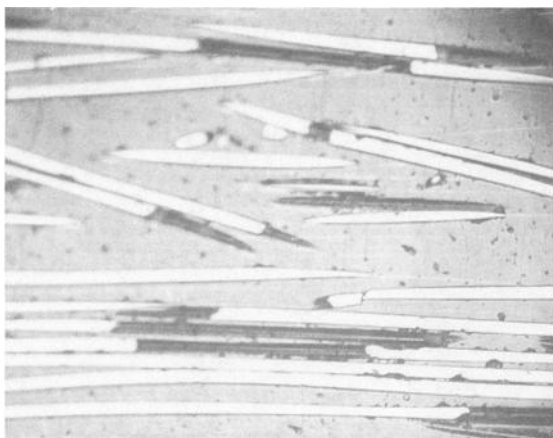


FIG. 12—Photomicrograph illustrating fiber misalignment in unidirectional SiC/CAS composite.

linear behavior corresponding to matrix cracking in the transverse to the fiber direction and to debonding at the interface between fiber and matrix as a result of the high interfacial shear stress concentration. The prediction for this stress is 214 MPa (31 ksi). Region BC is believed to correspond to transverse matrix crack multiplication up to a saturation crack density to be discussed below. Fiber fractures, which may start before transverse matrix crack saturation, and debonding continue at least through Region CD. The prediction in the Region CD shows a marked difference from experimental results, which is attributed to the following factors: (1) the orientation of the fibers of the tested specimen was not uniform so that the loading direction was not aligned perfectly with the fiber direction (Fig. 12) and shear deformation at the interface increases the apparent average axial strain of the specimen; (2) the interfacial shear strength and thermal residual stresses, which affect very much the stress distribution at the interface between matrix cracks, are not known; (3) the frictional effect at the debonded interface was neglected; and (4) earlier fiber fractures were not considered in the prediction. Beyond Point D the material behaves linearly to failure with decreasing secant modulus (Region DE), which corresponds to continuing fiber fractures with the same damage rate and small increase of debonded length. The terminal tangent modulus obtained from the test is $E_t = 47$ GPa (6.8 Msi); the predicted one is 70 GPa (10.2 Msi) where the fiber breaks were not considered. At the final failure of the specimen (Point E), the predicted failure strength from this analysis, 814 MPa (118 ksi), is almost twice as high as the experimental one. The difference may be due to earlier fiber fractures, thermal residual stresses, and fiber misalignment.

In Figure 13, transverse matrix crack density was predicted using this analysis and compared with the measured data from the microscope along the surface of the specimen [1]. The predicted transverse crack density increases rapidly between 211 and 221 MPa (30.6 and 32 ksi), then it reaches a saturation level of 29.3 cracks/mm or a minimum crack spacing of 34 μ m (0.00134 in.). However, the experimental results show earlier transverse matrix cracking at an applied stress of 110 MPa (16 ksi) increasing rapidly between 180 and 318 MPa (26 and 46 ksi). This difference is due to the statistical tensile strength variation of the matrix material. The saturation crack spacing measured was 28 cracks/mm, which is very close to the predicted value.

The ratio of predicted debonded length to the crack spacing was plotted as a function of applied stress in Fig. 13. The debonded length ratio increases rapidly up to 75% of the crack spacing between applied stresses of 221 and 262 MPa (32 and 38 ksi), then it increases grad-

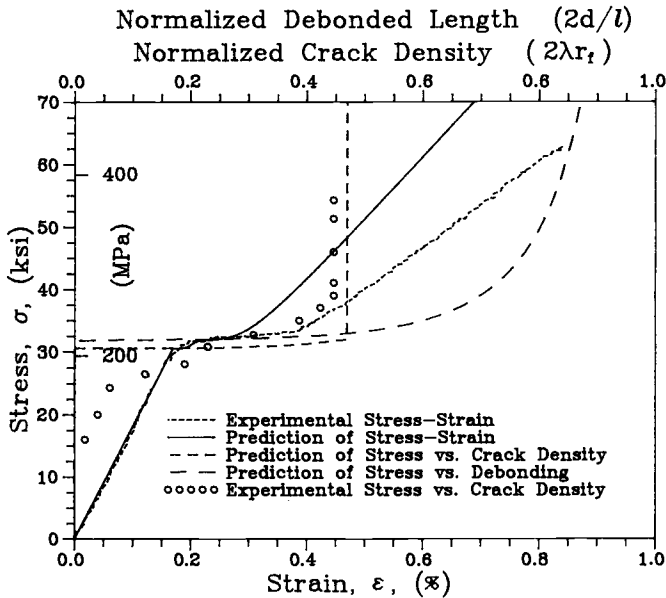


FIG. 13—Stress-strain, stress versus matrix crack density and stress versus debonded length curves for SiC/CAS [0₈] unidirectional composite.

ually until failure. It was not possible to observe the exact debonded length at the interface under the microscope, so this prediction cannot be compared with experimental results.

Summary and Conclusions

A modified shear lag analysis was conducted for a unidirectional brittle-matrix composite under longitudinal tensile loading. It is assumed that the matrix, having a lower ultimate strain than the fiber, fails by transverse cracking. The analysis was conducted for a cylindrical element of matrix with a single fiber for various boundary conditions. The shear lag parameter is expressed only as a function of geometry and shear moduli of the constituents. The basic assumptions used in the analysis are that the shear stress in the fiber varies linearly and in the matrix varies according to an inverse function of a second-degree polynomial in the radial direction. The analysis can take into account residual stresses in the material. Closed form solutions for stress distributions in the matrix and fiber and at the interface were obtained and compared with other available analytical results. They agree well with each other.

The transverse matrix crack density, debonded length, and reduced axial stiffness of the damaged model with two matrix cracks were obtained in closed form as a function of applied stress, constituent properties, and given damage state.

Stress-strain, stress versus matrix crack density, and stress versus debonded length curves were obtained. The failure mechanisms were studied under the microscope for a unidirectional silicon carbide/glass-ceramic composite loaded under longitudinal tensile loading. Matrix cracking was observed increasing in density up to a limiting level of approximately 28 cracks/mm or a minimum crack spacing of 36 μm, while the prediction gives 29 cracks/mm or a minimum crack spacing of 34 μm. Evidence of fiber debonding and fiber fractures was also observed under the microscope and in the measured stress-strain curve. Predicted results for stress-strain and stress versus matrix crack density curves are in good qualitative agreement

with experimental results. However, the present model is preliminary. A more comprehensive model is expected in the future by considering the frictional effect in the debonded area and using more realistic fracture criteria and constituent properties.

Acknowledgment

The work described here was sponsored by the Air Force Office of Scientific Research (AFOSR). We are grateful to Lt. Col. George Hantos of the AFOSR for his encouragement and cooperation, to G. Anastassopoulos for his contribution of experimental data, and to Yolande Mallian for typing the manuscript.

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A Macro–Micromechanics Analysis of a Notched Metal Matrix Composite

REFERENCE: Bigelow, C. A. and Naik, R. A., “A Macro–Micromechanics Analysis of a Notched Metal Matrix Composite,” *Composite Materials: Testing and Design (Tenth Volume)*, ASTM STP 1120, Glenn C. Grimes, Ed., American Society for Testing and Materials, Philadelphia, 1992, pp. 222–233

ABSTRACT: A macro–micromechanics analysis was formulated to determine the matrix and fiber behavior near the notch tip in a center-notched metal matrix composite. Results are presented for a boron/aluminum monolayer. The macro-level (or global) analysis models the entire notched specimen using a three-dimensional finite element program that uses the vanishing-fiber-diameter model to model the elastic-plastic behavior of the matrix and the elastic behavior of the fiber. The micro-behavior is analyzed using a discrete fiber-matrix (DFM) model containing one fiber and the surrounding matrix. The dimensions of the DFM model were determined by the ply thickness and the fiber volume fraction and corresponded to the size of the notch-tip element in the macro-level analysis. The boundary conditions applied to the DFM model were determined from the macro-level analysis. Stress components within the DFM model were calculated, and stress distributions are presented along selected planes and surfaces within the DFM model, including the fiber-matrix interface. Yielding in the matrix was examined at the notch tip in both the macro- and micro-level analyses. The DFM model predicted higher stresses (24%) in the fiber than predicted by the global analysis. In the notch-tip element, the interface stresses indicated that a multi-axial stress criterion may be required to predict interfacial failure. The DFM analysis predicted yielding to initiate in the notch-tip element at a stress level 28% lower than predicted by the global analysis.

KEYWORDS: yielding, monolayer, boron/aluminum, interface, stress contours

Metal matrix composites have several inherent properties, such as high stiffness-to-weight ratios and high strength-to-weight ratios, that make them attractive for structural applications. These composites also have a higher operating temperature range and better environmental resistance than current polymer matrix composites. Like polymer matrix composites, however, metal matrix composites are notch-sensitive. Unlike typical polymer matrix composites, metal matrix composites may exhibit widespread yielding of the matrix before laminate failure. To design damage-tolerant structures (or simply to understand the effects of fastener holes), the laminate fracture strengths must be known for a wide range of ply orientations, notch geometries, and loading conditions. A method for predicting fracture strength is needed to avoid testing all the laminate, notch, and loading combinations of interest.

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The present paper combines a three-dimensional homogeneous, orthotropic finite element analysis of a center-notched tension specimen and a discrete fiber-matrix micromechanics model of a single fiber. The displacements predicted by the macro-level analysis of the tensile specimen were imposed as boundary conditions on the discrete fiber-matrix (DFM) micro-model. To demonstrate this approach, results are presented for a single metal matrix composite, a boron/aluminum (B/Al) monolayer. Testing of this specimen was described in Ref 1.

A global-local, or substructuring, approach has been used in the past for the analysis of complex structures, where the global analysis would address a large structure and the local analysis would analyze structural details, such as bolt holes (for example, Ref 2). In the present work, a slightly different approach is used in that the local analysis is a micromechanics analysis, analyzing a single fiber and its surrounding matrix as two discrete materials. Work has also been done using the "unit cell" micromechanical analysis, assuming an infinitely repeating array of fibers with the appropriate boundary conditions (see, for example, Refs 3-5). Square arrays, hexagonal arrays, and other configurations have been used. The present work uses the basic configuration of a unit cell, but in a slightly different manner. Instead of applying the boundary conditions corresponding to some assumed array of fibers and analyzing a unit cell representative of this array, the necessary displacement boundary conditions are obtained from the global or macro-level analysis of the notched specimen. By using these specific boundary conditions, the unit cell of fiber and matrix is in a stress state identical to that occurring in the macro-level analysis, thus analyzing a unit cell at a specific location in a laminate. The approach used in this work is referred to as a macro-micro analysis.

This approach, in effect, joins an element with discrete fiber and matrix to elements that are orthotropic but homogeneous. Obviously, for this to be valid, both types of elements must have equivalent stiffnesses. This has been shown to be true at the macro-level where both models predicted nearly identical material properties [6]. This approach also assumes that the boundary displacements predicted by the homogeneous element are not significantly different from those that would occur if a heterogeneous medium was accurately modeled. By using displacement boundary conditions, rather than stress boundary conditions, any errors introduced by this approach should be minimal.

This paper demonstrates the viability of this approach to a macro-micro analysis and presents analysis results for a unidirectional boron/aluminum monolayer with a center notch. This paper presents a brief description of the macro- and micro-level finite element programs. Next, the procedure used to link the macro-level and DFM models is described. Stress components within the DFM model were calculated, and stress distributions are presented along selected planes and surfaces within the DFM model, including the fiber-matrix interface. Finally, yielding in the matrix is examined near the notch tip.

Specimen and Geometry

The composite specimen modeled in the analyses was a boron/aluminum (B/Al) monolayer made by diffusion-bonding a sheet of 0.142-mm-diameter boron fibers between two 6061 aluminum foils. The monolayer was 0.279 mm thick with a fiber volume fraction of 0.30. The material was used in the as-fabricated condition, that is, it was not heat-treated. The monolayer specimen, as shown in Fig. 1, was 78 mm wide (2W), 101 mm long (2H), with a center cracklike notch 19 mm long (2a) and 0.127 mm wide. The nonlinear behavior of the matrix material was fit with a Ramberg-Osgood equation, where the modulus of the aluminum was 72 340 MPa, Poisson's ratio was 0.30, and the proportional limit was 34.5 MPa. The fiber was assumed to be a homogeneous, isotropic material with a modulus of 400 000 MPa and a Poisson's ratio of 0.13 [1]. Testing of this specimen is described in Ref 1.

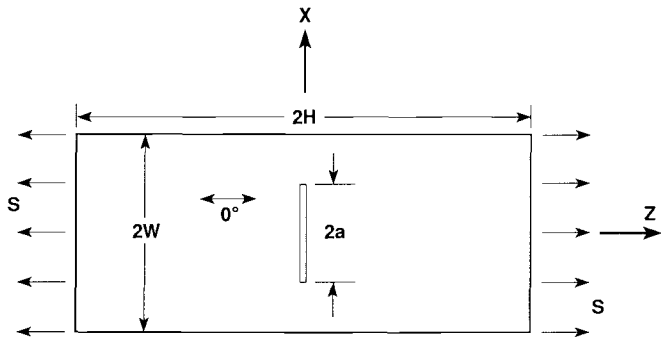


FIG. 1—Specimen geometry and loading. $2a = 19\text{ mm}$, $2W = 78\text{ mm}$, and $2H = 101\text{ mm}$, $\nu_f = 0.30$.

Analytical Modeling

Macro-Level Analysis

The macro-level analysis was conducted with a three-dimensional finite element program called PAFAC (Plastic and Failure Analysis of Composites) [7], which was developed from a program written by Bahei-El-Din et al. [8,9]. PAFAC uses a constant strain, eight-noded, hexahedral element. Each hexahedral element represents a unidirectional composite material whose fibers can be oriented in the appropriate direction in the structural (Cartesian) coordinate system. The PAFAC program uses the vanishing-fiber-diameter (VFD) material model developed by Bahei-El-Din and Dvorak [8,10,11] to model the elastic-plastic matrix and elastic fiber. The Ramberg-Osgood equation was used to model the nonlinear stress-strain curve of the aluminum matrix. The fiber was modeled as a linear, elastic material. The PAFAC program predicts fiber and matrix stresses at each element centroid.

The finite element mesh used to model the center-notched monolayer is shown in Fig. 2. Only one eighth of the specimen was modeled because of symmetry. The macro-level mesh is

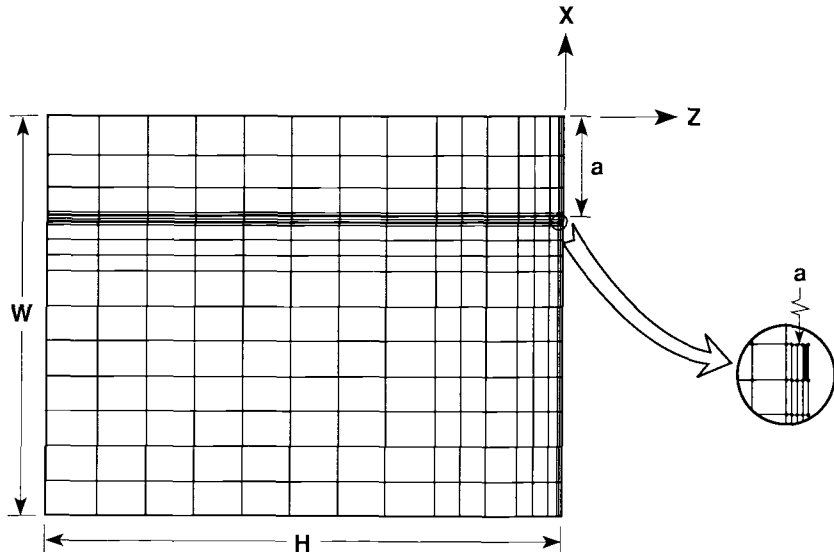


FIG. 2—Macro-level finite element mesh.

defined with respect to the global x -, y -, and z -axes as indicated in Fig. 2. This mesh used 1389 nodes and 800 elements. As shown in Fig. 1, a uniform stress in the z -direction was applied to the end of the specimen to simulate tensile loading. The notch-tip element modeled by the micro-analysis is indicated in Fig. 2 by the shaded element. The macro-analysis predicted the maximum fiber axial stress in the element at this location.

In the macro-analysis, further mesh refinement could have been used. However, for an individual element to have any physical meaning, the smallest size possible is that of a single fiber and its surrounding matrix. Thus, the notch-tip element was sized to represent one fiber and the surrounding matrix. The dimensions of this element were calculated based on the fiber diameter, monolayer thickness, and fiber volume fraction of the composite. The rectangular mesh pattern shown in Fig. 2 was chosen to model the longitudinal shear yielding at the notch tip in unidirectional laminates [12].

Discrete Fiber-Matrix Model

Figure 3 shows how the DFM model was defined with respect to the macro-level mesh and where it is located with respect to the notch tip. Because of symmetry, one half of the fiber and the surrounding matrix were modeled. Figure 4 shows the dimensions of the DFM model (it has the same dimensions as the notch-tip element in the macro-level analysis). Figure 4 shows the finite element mesh used in the DFM model. The DFM mesh is defined with respect to the local x -, y -, and z -axes as indicated in Fig. 4. A convergence study was performed to determine the mesh refinement by examining the radial and tangential stresses along the fiber-matrix interface for various mesh refinements. Eight layers of elements were used in the fiber or z -direction. The MSC/NASTRAN finite element code was used in the DFM analysis with eight-noded, isoparametric, hexahedral elements. A piece-wise linear approximation of the Ramberg-Osgood equation was used in NASTRAN to model the nonlinear stress-strain curve of the aluminum. The fiber was modeled as a linear, elastic material.

Determination of DFM Boundary Conditions

As mentioned earlier, the boundary conditions applied to the DFM model were determined from the macro-level finite element analysis. Displacement boundary conditions were used. The displacements of the element at the notch tip were used to calculate the boundary conditions to be applied to the DFM model. The displacements for the corner nodes in the DFM model were taken directly from the macro-level analysis. To determine the displacements for the nodes on all the surfaces of the DFM model, second-order Lagrangian interpolation was used assuming that the nodal displacements from the macro-level analysis were a bivariate function. The displacements in the thickness or y -direction were assumed to vary linearly.

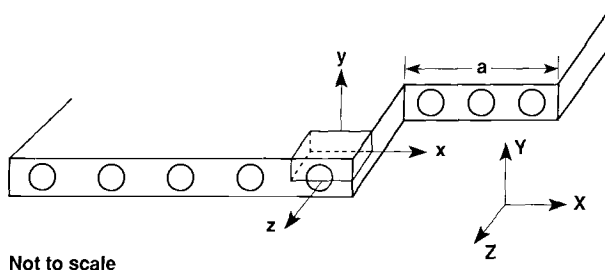


FIG. 3 Definition of micro- and macro-level models.

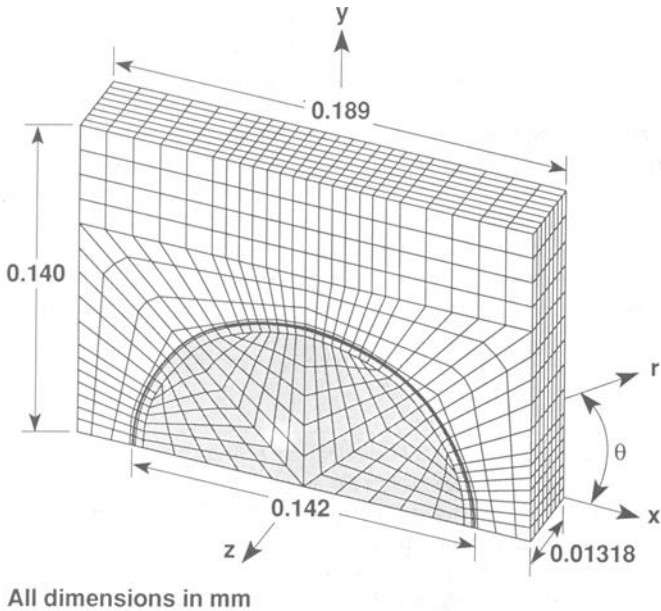


FIG. 4—Discrete fiber-matrix (DFM) model.

Because of symmetry in the macro-level model, the y -displacements on the local $y = 0$ face and the z -displacements on the $z = 0.0318$ -mm face were constrained to be zero. To illustrate the calculation of the other boundary conditions, consider the geometry defined in Fig. 5. The element drawn in solid lines is the element at the notch tip modeled by the DFM model. The elements drawn in dashed lines are the adjacent three-dimensional elements in the macro-level analysis. The displacements from adjacent nodes were used to determine the appropriate boundary conditions as follows. Consider, for example, the face of the DFM model in the xz -plane defined by the node Points 6, 8, 16, and 14 in Fig. 5. The displacements on the larger surface in xz -plane defined by Nodes 2, 8, 32, and 26 were fit to a bivariate function using a second-order polynomial to determine an equation for the displacements of this surface. The displacements of all 16 nodes on this surface were used to determine this surface equation. The coordinates of all nodes on this face in the DFM model were used with this equation to

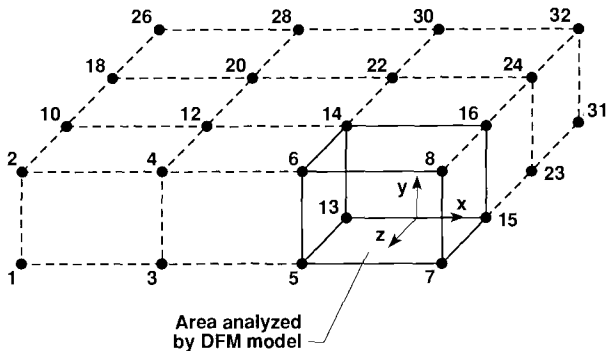


FIG. 5—Macro-level nodes used to define DFM boundary conditions.

calculate the displacements of the boundary nodes of the DFM model. Because the macro-analysis model had only one layer of elements in the y -direction, macro-level nodal displacements are available for only two node points, and a linear displacement distribution was assumed through the thickness.

A set of displacement boundary conditions were determined for the element location in the macro-level model shown in Fig. 3, the element next to the notch tip.

Results and Discussion

In all cases, results are presented for element centroidal stresses. Stress contours for the fiber and matrix, and plots of the interface stresses are presented for a unit applied remote stress. Stresses are presented with respect to the cylindrical coordinate system defined in Fig. 4. Yielding predicted by the DFM analysis is also presented.

Matrix Stresses

Stress contours are presented for an elastic stress state due to a remotely applied unit stress ($S = 1.0$ MPa in Fig. 1). Figures 6, 7, and 8 show predicted stress contours for the $z = 0.0318$ mm face of the DFM model. The notch tip is located on the right side of this face in the DFM model in the figures. These figures show the matrix only in the DFM model. For clarity, fiber stresses are shown separately in a following section.

Figure 6 presents the σ_{rr} stress contours in the matrix. The highest σ_{rr} stresses are on the side of the fiber away from the notch tip at $\theta = 180^\circ$ due to the constraint provided by the adjacent fiber. However, the stresses next to the notch ($\theta = 0^\circ$) are nearly as large as at $\theta = 180^\circ$. Some small compressive stresses are present next to the fiber-matrix interface near $\theta = 100^\circ$.

Figure 7 present the $\sigma_{\theta\theta}$ stress contours in the matrix. These stresses are of approximately the same magnitude as the σ_{rr} stresses. The maximum $\sigma_{\theta\theta}$ stresses occur at the free surface at the upper corner next to the notch tip. The $\sigma_{\theta\theta}$ stresses around the fiber are greatest on the side of the fiber next to the notch tip.

Figure 8 presents the σ_{zz} stress contours in the matrix. This is the stress component in the fiber axial and loading direction; thus, these stresses are of the largest magnitude. As expected, the largest σ_{zz} stresses occur at the free surface next to the notch tip near $\theta = 0^\circ$ and predict a matrix stress concentration factor greater than 7. The macro-analysis predicted a matrix stress

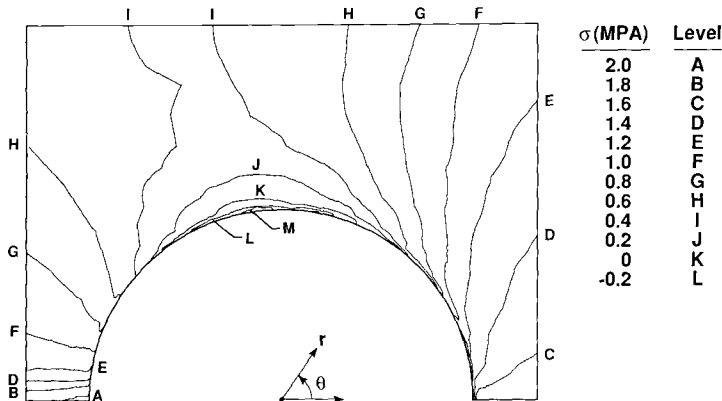


FIG. 6 Matrix radial stress contours in notch tip element for unit applied remote stress ($S = 1.0$ MPa).

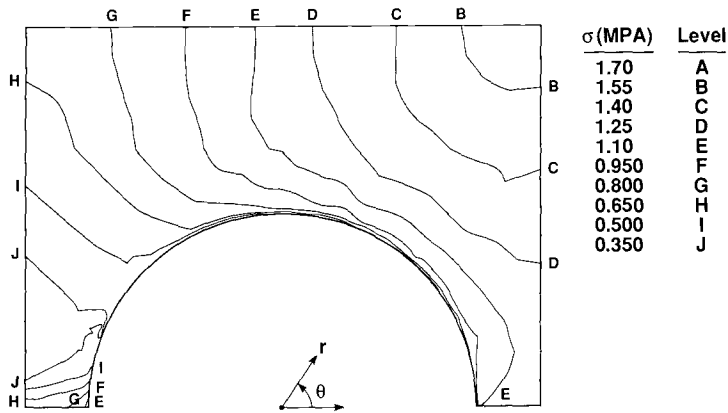


FIG. 7—Matrix $\sigma_{\theta\theta}$ stress contours in notch-tip element for unit applied remote stress ($S = 1.0$ MPa).

concentration factor of 11 for the remote unit stress. In the DFM model, the σ_{zz} stresses increase fairly uniformly across the width with increasing distance from the notch. A second stress concentration occurs at $\theta = 180^\circ$ as a result of the constraint provided by the adjacent fiber.

The $\tau_{r\theta}$ stresses are of particular interest at the fiber-matrix interface (this stress component is discussed in the next section). Because of symmetry, the τ_{rz} and $\tau_{\theta z}$ are zero on the xy -plane ($z = 0.318$ -mm face); stress contours are not shown for these components.

Interface Stresses

Because material properties for the interface between the fiber and the matrix are not known, the interface region was not modeled as a discrete region with distinct properties; only fiber and matrix were modeled. Interface stresses were calculated in the layer of matrix elements next to the fiber. Interface stresses are presented with respect to the cylindrical coordinate system shown in Fig. 4, where $\theta = 0^\circ$ is the side of the fiber closest to the notch tip. Interface stresses are shown for an elastic stress state caused by a remotely applied unit stress ($S = 1.0$ MPa in Fig. 1).

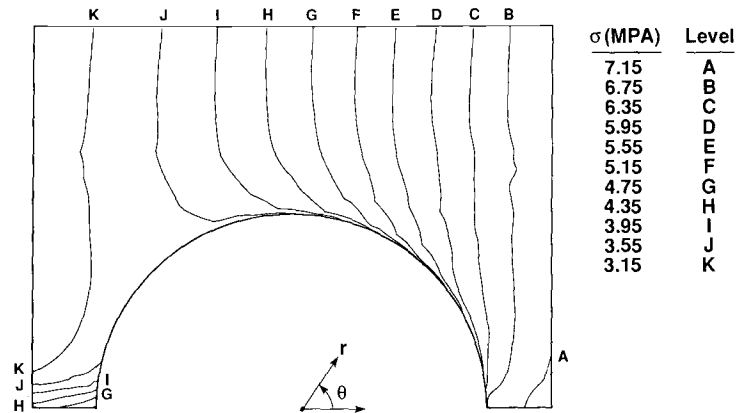


FIG. 8—Matrix σ_{zz} stress contours in notch-tip element for unit applied remote stress ($S = 1.0$ MPa).

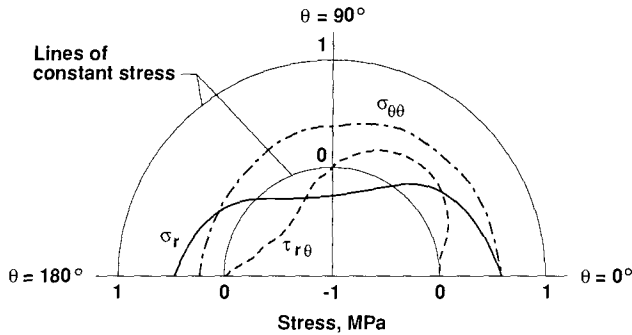


FIG. 9—Interface stresses for unit applied remote stress ($S = 1.0$ MPa).

Figure 9 presents a polar plot of the interface stresses due to a remote unit stress. The σ_{rr} stresses (solid curve) have a maximum tensile value at $\theta = 0^\circ$ and 180° , with the slightly larger stress at $\theta = 0^\circ$, next to the notch tip. The σ_{rr} stress has a compressive peak at $\theta = 85^\circ$. Due to the σ_{rr} component, the matrix wants to pull away from the fiber along the specimen midplane, while compressing the fiber near $\theta = 90^\circ$.

Because of symmetry, $\tau_{r\theta}$ (dashed curve) is zero at $\theta = 0^\circ$ and 180° . The maximum values of $\tau_{r\theta}$ occur near $\theta = 140^\circ$ (negative shear stress) and $\theta = 50^\circ$ (positive shear stress).

In general, the $\tau_{r\theta}$ shear stress should be considered as an interface stress. However, in this case, as a result of symmetry, this stress component is zero on the $z = 0.0318$ -mm face of the DFM model.

Figure 9 also shows the $\sigma_{\theta\theta}$ stress at the interface (dash-dot curve). While this stress component is not usually considered an interface stress because it does not act in the plane of the interface, it is an important component to consider in determining fiber-matrix separation. A tensile or positive $\sigma_{\theta\theta}$ stress causes the matrix to grip the fiber, even if the interface has completely failed. This behavior was observed experimentally for SCS₆/Ti-15-3 composites [13]. Figure 9 shows that, for the notch-tip element, $\sigma_{\theta\theta}$ is tensile for all values of θ and reaches a maximum value at $\theta = 0^\circ$.

Along the specimen midplane ($\theta = 0^\circ$ and 180°), the σ_{rr} stress would cause the matrix to pull away from the fiber, despite the tensile $\sigma_{\theta\theta}$ stresses that generally cause the matrix to grip the fiber. If σ_{rr} governs failure of the interface, the most likely location for failure is at $\theta = 0^\circ$. Normal and shear components vary around the fiber, suggesting that a multi-axial stress criterion may be required to predict interfacial failure and fiber-matrix separation. One caution should be noted. This analysis does not account for thermal residual stresses. Predictions of the stress state at the interface can be significantly altered by including thermal residual stresses [13].

Fiber Stresses

The fiber stresses are presented for an elastic stress state caused by a remotely applied unit stress ($S = 1.0$ MPa in Fig. 1). Figures 10, 11, and 12 show predicted stress contours for the $z = 0.0318$ -mm face of the DFM model. For clarity, only the fiber is shown in these figures. The notch tip is located to the right side of the fiber in these figures.

Figures 10 and 11 show the σ_{rr} and $\sigma_{\theta\theta}$ stress contours, respectively, for the fiber. Figure 10 shows that the σ_{rr} stresses are compressive through most of the fiber, becoming tensile on the side of the fiber away from the crack. The largest magnitude σ_{rr} stresses occur at $\theta = 90^\circ$ near the fiber centroid. The $\sigma_{\theta\theta}$ stresses, shown in Fig. 11, are entirely compressive with the largest

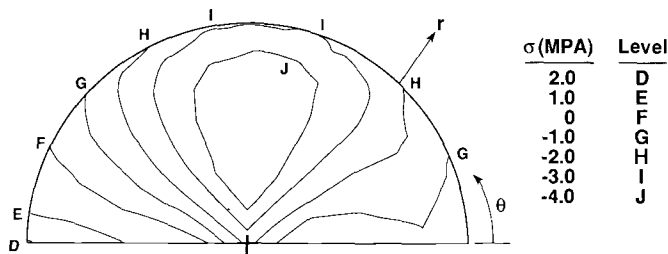


FIG. 10—Fiber σ_{π} stress contours in notch-tip element for unit applied remote stress ($S = 1.0$ MPa).

The macro-level analysis calculated the σ_{zz} stress in the fiber in the notch-tip element to be 25 MPa for a remote unit applied stress. Figure 12 presents the σ_{zz} stress contours in the fiber for the remote unit stress. As expected, the σ_{zz} stresses are largest on the side of the fiber next to the notch tip. The maximum point value of the σ_{zz} stress is 32 MPa at $\theta = 0^\circ$ with an average σ_{zz} stress of 22 MPa, compared to 25 MPa predicted by the macro-analysis. Similar differences between the DFM and macro-analysis predictions were found for the other fiber stress components. This is expected because the DFM analysis calculates a varying stress distribution through the cross section, whereas the macro-analysis only calculates a single value for each stress component. Thus, the two analyses agree very well for average values, but a homogeneous model such as the VFD cannot model such detailed behavior as is possible with the DFM analysis.

Yielding

Based on the von Mises yield criterion, the macro-level analysis predicted yielding of the notch-tip element at an applied remote stress of 8 MPa. A nonlinear analysis using the DFM model was done for the displacement boundary conditions corresponding to this remote stress. Figure 13 shows the von Mises equivalent stress contours for $S = 8$ MPa, where the von Mises equivalent stress, $\bar{\sigma}_{vm}$, is defined as follows:

$$\bar{\sigma}_{vm} = \sqrt{\sigma_x^2 + \sigma_y^2 + \sigma_z^2 - \sigma_x\sigma_y - \sigma_y\sigma_z - \sigma_z\sigma_x + 3(\tau_{xy}^2 + \tau_{yz}^2 + \tau_{zx}^2)}$$

When the von Mises equivalent stress is greater than or equal to the proportional limit ($\bar{\sigma}_{vm} \geq 34.5$ MPa), the matrix elements within the DFM mesh have yielded. Thus, in Fig. 13, the area to the right of contour line A has yielded; approximately one third of the DFM model has yielded at the remote stress of 8.0 MPa. The macro-analysis predicted that the entire element would yield at this stress level.

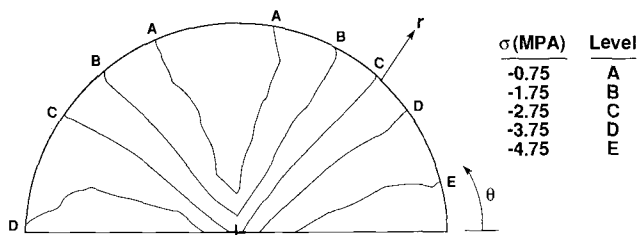


FIG. 13—Fiber σ_{π} stress contours in notch-tip element for unit applied remote stress ($S = 1.0$ MPa).

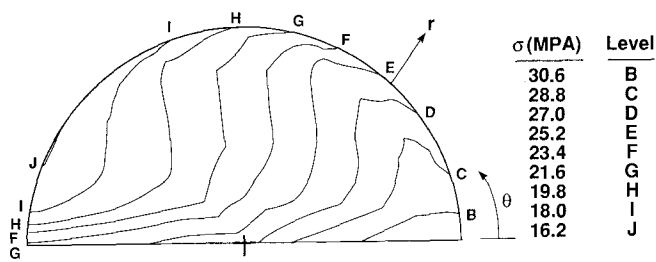


FIG. 12—Fiber σ_{zz} stress contours in notch-tip element for unit applied remote stress ($S = 1.0$ MPa).

The DFM analysis predicted that yielding initiated next to the notch tip at an applied remote stress of 5.75 MPa, a value 28% lower than the stress level of 8.0 MPa predicted by the macro-level analysis. As with the stress predictions, this difference is not unexpected. The finer mesh used in the micro-analysis results in predictions of steeper stress gradients and higher peak stresses, which leads to predictions of lower remote stresses to yield the matrix. Both analyses agreed when predicting the unnotched laminate stress-strain curve.

All the displacement boundary conditions calculated from the macro-analysis are for an elastic stress state. As mentioned earlier, these displacement boundary conditions are valid only if any yielding in the DFM model is limited to a small region. Large-scale yielding in the DFM model would affect the displacements of the boundary nodes; thus, the displacements predicted by the macro-level analysis assuming that the notch-tip element is behaving elastically would no longer be valid. Figure 11 shows that the yielding is indeed small-scale; therefore, the displacement boundary conditions from the macro-analysis are valid.

Figure 14 shows a plot of the von Mises equivalent stresses in the fiber for the remote stress $S = 8.0$ MPa. Because the fiber is assumed to behave elastically, these stresses have no meaning in relation to a yield criterion. However, this equivalent stress can be viewed in terms of a failure criterion for the fiber. The maximum equivalent stress is predicted at the edge of the fiber next to the notch tip. Because the matrix and the interface in this region are also highly stressed, this area would be a likely location for crack initiation and interface failure. In this particular case, as a result of symmetry, the equivalent fiber stresses for the notch-tip element are very similar to the fiber σ_{zz} stresses shown in Fig. 12, thus, σ_{zz} alone could be used to predict fiber failures. However, in general, all stress components should be considered for a fiber fail-

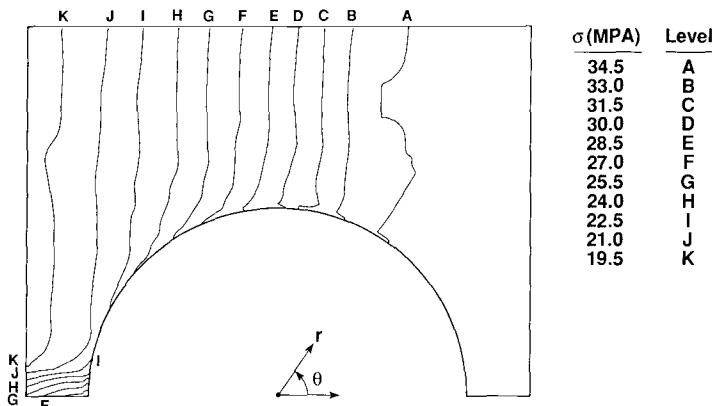


FIG. 13—Von Mises equivalent stress contours in matrix for applied remote stress ($S = 8.0$ MPa).

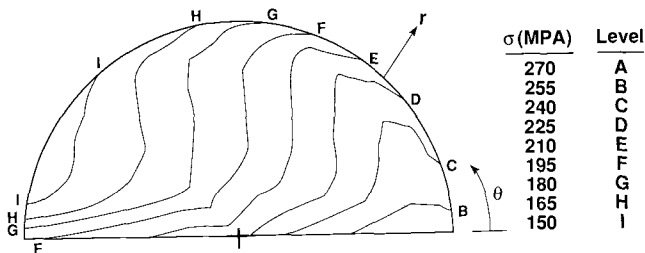


FIG. 14—*Von Mises equivalent stress contours in fiber for applied remote stress ($S = 8.0$ MPa).*

ure criterion. Earlier work with notched B/Al laminates [1] used a two-parameter failure criterion based on axial and shear stress in the fiber to accurately predict the first fiber failure.

Conclusions

A macro–micromechanics analysis was formulated to determine the matrix and fiber behavior near the notch tip in a center-notched metal matrix composite. The viability of this approach is demonstrated, and results are presented for a boron/aluminum monolayer. The macro-level analysis models the entire notched specimen using a three-dimensional homogeneous, orthotropic finite element program that uses the vanishing-fiber-diameter (VFD) model to compute the elastic-plastic behavior of the matrix and the elastic behavior of the fiber. The micro-behavior is analyzed using a discrete fiber-matrix (DFM) model containing one fiber and the surrounding matrix. The boundary conditions applied to the DFM model were determined from the ply thickness and the fiber volume fraction and corresponded to the size of the notch-tip element in the macro-level analysis. Stress components within the DFM model were calculated, and stress distributions are presented along selected planes and surfaces within the DFM model, including the fiber-matrix interface.

The DFM model predicted significantly higher maximum stresses in the notch-tip matrix and fiber, where a very nonuniform stress state was present, compared to the macro-analysis, which assumes an average stress state in each element. In particular, the predictions of fiber axial stress differed by 24%. However, the two analyses agree very well when predicting gross behavior such as moduli or unnotched stress-strain behavior.

In the notch-tip element, the interface stresses had significant shear and normal components, indicating that a multi-axial stress criterion may be necessary to analyze and predict fiber-matrix interface failure and separation. A micro-level analysis, such as the DFM model, is necessary to analyze and predict interfacial behavior.

The DFM analysis predicted that yielding initiated at the free surface next to the notch tip and progressed rather uniformly through the notch-tip element in a direction away from the notch tip. The DFM analysis predicted the initiation of yielding at a much lower stress level than the macro-analysis (28% difference), but the macro-analysis predicted the notch-tip element to be completely yielded at a stress level where the DFM analysis predicted only partial yielding.

To analyze and predict interface stresses or stress distributions through a fiber cross section accurately, a micro-level analysis is required. A global analysis, such as the VFD model, is accurate only for predicting behavior in an average sense. When combined, the two analyses can provide a detailed understanding of the microdamage development in metal matrix

composites

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Session V

Analysis, Test, and Certification of Structure

Thomas J. Dunyak, Chairman

Overview

One of the major challenges facing the composites community is the extent to which we can make a transition from the properties of materials to the application of composites to safe, reliable structures. This transition is a very complex process which includes design of the structure, analysis of the design, experimental verification of the analysis, and final certification of the structure. Each of the four papers presented in this session deal with one or more aspects of this transition process.

The first paper in this session by Paul, Saff, Sanger, et al. describes the "Analysis and Test Techniques for Composite Structures Subjected to Out-of-Plane Loads." The first part of the paper contains a summary of several analytical packages developed for the preliminary analysis of laminate corner radii stresses, flange-to-skin bondline stresses, and stiffener runout stresses. The second part of the paper describes relatively simple test methods for the verification of these analyses.

The next paper entitled "Analysis and Testing of a Composite Sandwich Shell Horizontal Tail" by Sherrouse and McWhorter investigates the feasibility of an all-composite horizontal tail for light aircraft. This type of structure offers many advantages over conventional tail structures by simplifying construction, reducing weight, and requiring far fewer fasteners. The paper describes the design, analysis, fabrication, and experimental testing of the structure subjected to various point loads. The final section compares the analytical and experimental results.

The third paper entitled, "Composite Material Stub-Blade Wing Joint" by Franklin and Kreimendahl describes the design of an all-composite stub-blade joint for high-altitude, long-endurance aircraft which incorporate removable, composite wing structures with large spans and high aspect ratios. The main body of the paper is divided into three major sections. The first section describes the design philosophy including design requirements, possible design alternatives, and justification of the final design selection. The second section presents a detailed description and analysis of the final design, and the final section compares the all-composite blade joint to a traditional, titanium tension fitting.

The final paper in this session entitled, "Temperature and Load Cycling of a Thick Polyimide Quartz Laminate" by Laurencot and Keen describes the temperature and out-of-plane load cycling on a thick polyimide/quartz V-shaped sub-element. The paper presents detailed descriptions of the experimental test setup, applied load history, and measured response of the material.

All of these papers describe the application of composite materials to complex engineering structures. In some of the papers, the authors investigate the design of all composite structures, and in other cases, the authors describe highly specialized experimental techniques based on well-established ASTM procedures. Overall, each of these papers exemplify a portion of this transition from properties of laboratory coupons to complex, yet reliable, engineering structures.

Philip C. Paul,¹ Charles R. Saff,² Kenneth B. Sanger,³ Mary A. Mahler,⁴ Han-Pin Kan,⁴ and Edward F. Kautz⁵

Analysis and Test Techniques for Composite Structures Subjected to Out-of-Plane Loads

REFERENCE. Paul, P. C., Saff, C. R., Sanger, K. B., Mahler, M. A., Kan, H.-P., and Kautz, E. F., "Analysis and Test Techniques for Composite Structures Subjected to Out-of-Plane Loads," *Composite Materials: Testing and Design (Tenth Volume)*, ASTM STP 1120, Glenn C. Grimes, Ed., American Society for Testing and Materials, Philadelphia, 1992, pp. 238–252.

ABSTRACT: The ability to predict stresses and failures caused by out-of-plane loads has gained importance as airframe manufacturers begin to use integral (co-cured or bonded) composite structures to minimize weight and maximize performance. Rapid and accurate analysis methods are needed to reduce the amount of testing required to ensure confidence in integral composite structures. Recently, simple two-dimensional analysis methods were developed to predict the out-of-plane failure strengths of composite airframe structures. These analyses were primarily derived to address induced stresses in laminate corner radii, direct stresses as a result of fuel pressure loads, induced stresses as a result of panel buckling, and induced stresses as a result of stiffener runouts or other load path changes.

The development and verification of analyses for out-of-plane failures have shown the need for out-of-plane material properties and strength allowables that are not traditionally determined for composite laminates. Accurate stress and failure predictions depend heavily on these experimentally determined values. The test methods and experience gained in this program indicate the need for new techniques to determine the interlaminar properties required for accurate stress and failure predictions. Test techniques are suggested for determining some of the values necessary for these predictions.

KEYWORDS: composite, composite structures, graphite/epoxy, out-of-plane analysis, co-cured, adhesively bonded, integral composite structure, interlaminar stress, translaminar stress, ply drop off, direct load, indirect load, delamination

The ability to predict the failure of co-cured and bonded joints as a result of out-of-plane loads is becoming increasingly important as airframe manufacturers begin to use integral composite structures to minimize structural weight and maximize airframe performance. Recent failures of composite airframe structures have demonstrated the difficulties encountered in predicting out-of-plane loads and the effect of such loads on strength of laminated, co-cured, or adhesively bonded composite structures. These unexpected failures have been directly or indirectly caused by translaminar (out-of-plane) loads.

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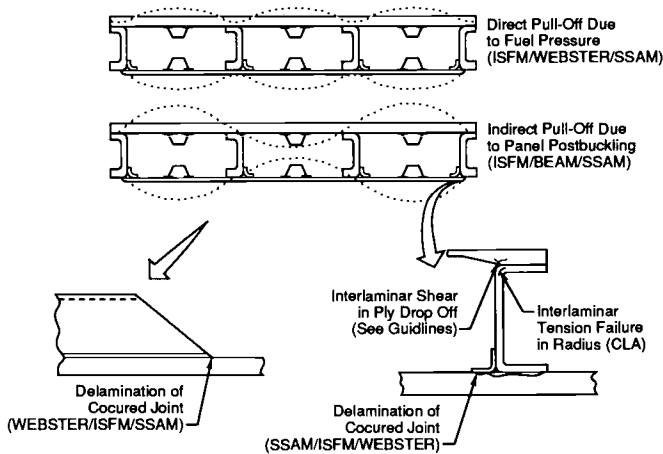


FIG. 1—Potential out-of-plane failures in integrally stiffened torque box design and recommended analysis routines.

The out-of-plane failures in composite structures result from the low strength of co-cured or bonded joints, integral stiffening, and the inadequate design tools available for determining stress concentrations caused by out-of-plane loads. Because of the inherently poor transverse strength of composite laminates, assuming no transverse reinforcement, laminated or bonded composite structures are prone to delamination under local out-of-plane loads. Analysis tools are needed to predict where local transverse strength is required to produce strong, durable, integral composite structures. Furthermore, out-of-plane loads must be considered in the design, validation, and certification phases of airframe development.

Analysis methods that predict out-of-plane failure modes and strengths, based on simple element tests, are needed to reduce the amount of testing required to ensure confidence in integral composite structures. While three-dimensional finite element analyses (FEA) can be used to analyze such structures, this modeling is too time-consuming to be used in preliminary design.

Simple two-dimensional analysis methods have been developed to predict the primary out-of-plane failure modes and strengths of composite airframe structures [1]. The primary sources of out-of-plane failures addressed by these analysis techniques are:

- induced stresses in laminate corner radii,
- direct stresses caused by fuel pressure loads,
- induced stresses caused by stiffener runouts or load path changes, and
- induced stresses caused by panel buckling.

Figure 1 identifies the potential out-of-plane failures of a typical integral composite torque box design. The recommended analysis routines for each failure are also identified [1].

McDonnell Aircraft Company (MCAIR) and Northrop Aircraft Division (NAD) teamed to perform a four-task program to develop out-of-plane analyses for composite structures. These four tasks are identified below:

- (1) perform a field survey to identify the most troublesome out-of-plane failure modes experienced in composite airframes;

- (2) develop strength analysis procedures for certain out-of-plane loading problems identified in the field survey,
- (3) verify these analysis procedures using test data from MCAIR and Northrop Internal and Contracted Research and Development Reports (IRAD and CRAD) and open literature, and
- (4) develop design guidelines and demonstrate the application of the analysis techniques to a major aircraft component

The test verification of these out-of-plane analysis techniques is the primary subject of this paper. Problems associated with out-of-plane loading tests are addressed, and possible approaches to solving those problems are proposed.

Both MCAIR and Northrop supplied element and subcomponent data for analysis development and verification. However, during the course of this program, it became evident that the through-the-thickness properties (stiffnesses, as well as strengths) of composite laminates and co-cured and bonded joints are not known. The lack of suitable test techniques for measuring out-of-plane properties for composite structures limits analysis verification and, consequently, limits the application of integral composite structures.

To predict failures of out-of-plane loaded composite structures accurately, coupon, element, and built-up specimens are required. These test data can be used along with the analysis tools developed by Paul et al [1] to predict the out-of-plane strength of co-cured and bonded structural joints.

Analysis Methods

Five different analysis techniques were developed by Paul et al [1]. It is beyond the scope of this paper to describe each of these in detail. The best discussions of each are in Ref 1.

Most of the analyses developed by Paul et al [1] were derived primarily to examine pull-off failures of skin/stiffener joints. The curved laminate analysis (CLA) was developed as a closed form solution for the stresses and strains in radii found between stiffener uprights and flanges. It is a two-dimensional analysis. BEAM, the interlaminar stress field model (ISFM), and the simplified stiffener model analysis (SSAM) are progressively better analyses of the flange-to-skin bondline stresses. The BEAM analysis addresses flange free-end failures by analyzing the skin and flange as bonded beams. ISFM treats the flange as a beam on an elastic foundation so that stresses and failures at either the free end or beneath the upright of a stiffener can be examined. SSAM is a simple finite element model developed to allow a better analysis of the stiffener attachment geometry in prediction of stiffener pull-off failures. The web-stiffener termination elasticity routine (WEBSTER) incorporates an iterative elasticity solution to the stiffener or flange termination to skin bond.

CLA, BEAM, ISFM, WEBSTER, and SSAM can all be used to predict failure of stiffener/skin joints under direct pull-off loads. Each of these routines has its own application (as indicated in Fig. 1) based on its set of assumed geometry, loads, boundary conditions, and limitations. Selection of the analysis is determined by these conditions. CLA solely addresses curved laminates like those found at the intersection of the upright and the flange of a stiffener. BEAM is most appropriately used in cases where the flange end is the likely failure location. Stiffened skins that are stitched through the stiffener flanges to prevent significant stresses beneath the web provide an example of such a case. ISFM is most appropriate in cases in which failures can occur beneath a discontinuous web. WEBSTER is a more general method but requires the most input data to run. SSAM is a simple, two-dimensional, finite-element-based technique and is the most general method. It was developed under MCAIR IRAD funding.

very late in the course of this program. All of these routines, including SSAM, are available through the Naval Air Development Center (NADC), Warminster, PA.

Coupon Tests

Coupon tests are used to generate the material properties and strength allowables required to predict the strengths of structural components. The primary parameters required in these analyses are the interlaminar tensile and shear strengths. However, in this program, it was found that transverse and shear moduli also have significance in failure prediction. Out-of-plane failures in composite structures are generally tension, shear, or mixed tension and shear failures. Therefore, allowables are needed for interlaminar tension and shear stresses for out-of-plane designs. For mixed-mode failure analysis, a quadratic criterion can be used with reasonable reliability. Interlaminar tensile allowables can be determined from flatwise tension or curved beam tests. Interlaminar shear allowables can be determined from shear beam or other standard shear tests. The coupon tests most often performed in the past to generate these data were flat, circular interlaminar tension specimens (Fig. 2) and shear beam specimens (Fig. 3).

The interlaminar tension test specimens are cylindrical cores, called plugs, taken directly from the laminate. These plugs are bonded to metal rods and then pulled to failure in a tension test machine. Often, the bondline fails, rather than the laminate. In such cases, the test section is machined to produce a smaller central section to force the failure into the laminate. However, the stress concentration provided by this notch is not well characterized. In addition, machining, bonding, and machine alignment tolerances combine to produce significant scatter in the translaminar strength results.

The shear strength data used in this program were generated using the standard ASTM shear beam specimen and the ASTM Test for Apparent Interlaminar Shear Strength of Parallel Fiber Composites by Short Beam Method (D 2344). One drawback to this specimen is that it permits only a strength assessment. The routines developed by Paul et al. [1] require stiffness data, in addition to strength data, in both translaminar shear and tension. While short-lap shear specimens can be used to obtain adhesive shear modulus, a good specimen for interlaminar tensile modulus is not known. For the analyses performed in this program, in-plane transverse moduli were used to estimate the interlaminar moduli, but the accuracy of this estimation is not known.

Composites exhibit more strength scatter than metals. While in-plane properties of composites exhibit only slightly more scatter than metals, out-of-plane properties show considerably more scatter than in-plane properties. This is demonstrated by comparison of reduction

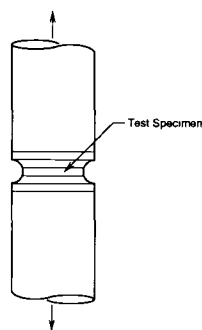


FIG 2—Interlaminar tension test specimen

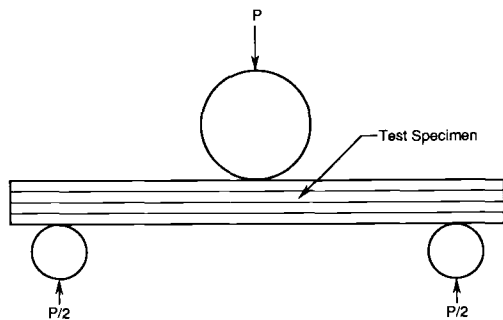


FIG. 3—Interlaminar shear test specimen.

factors for allowables under out-of-plane loadings versus in-plane loadings (Fig. 4). Because of this increased data scatter, an increase in the number of tests required to determine the B-basis design allowables from five to ten is suggested.

Out-of-plane properties are sensitive to environmental conditions. Therefore, design allowables should be determined for the entire range of the environmental service envelope of the structure. Statistical methods should be used to compute the design allowables as described in Refs 2 through 5.

In addition to environmental sensitivity, out-of-plane strengths can be reduced significantly by fatigue damage. Fatigue tests of flatwise tension coupons showed a rapid decrease in load-carrying capability even at very short lives (Fig. 5). On the other hand, fatigue endurance limit stresses (those below which no failures occur for extremely large numbers of cycles) of up to half of the ultimate interlaminar tensile strength have been found in test. Similarly, significant strength reductions have been found after fatigue loading of hat-stiffened panels (Fig. 6). The results shown in Figs. 5 and 6 suggest that one must consider repeated loads in developing allowables for out-of-plane loaded structures.

Element Tests

Element tests validate the out-of-plane analysis methodology and provide data for the determination of strength allowables. Two small elements are recommended for this purpose: curved laminate specimens and stiffener pull-off specimens.

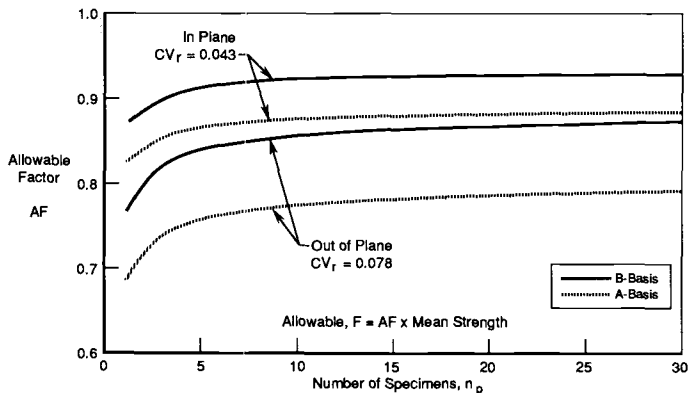


FIG. 4—The additional scatter in out-of-plane strengths is reflected in a reduced factor on allowables.

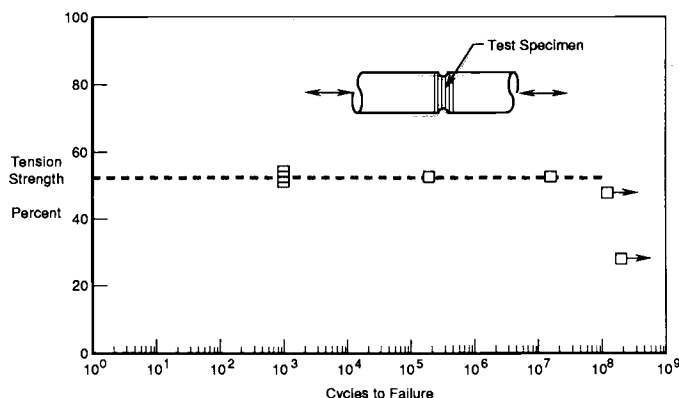


FIG. 5—Translaminar fatigue test results.

Curved Laminate Specimens—Curved laminates are used to attach spars and stiffeners to skins. The curved sections are used to redirect out-of-plane loads into the plane of the skin. The loads applied to curved laminates come from many sources: air pressure (as on the AV-8B blade strake), fuel pressure (skin to spar), and skin buckling (stiffener to skin). Once these loads are resolved into their normal and lateral load components and moments (Fig. 7), the CLA routine, a closed form analysis for stresses in curved composite laminates described in detail in Ref 1, can be used to determine the loads at which interlaminar shear or tensile failures occur.

Although the curved beam element specimen is more difficult to fabricate than the flatwise tension specimen, it is a preferred specimen for development of interlaminar tensile strength data. The primary concern about using the curved beam specimens to determine interlaminar strengths is that the failure occurs in a curved section that may not have the same degree of consolidation as the flat portion of the laminate. Unfortunately, the flat portion of the laminate is the area one most often wants to model.

The curved laminate test specimen and fixture are shown schematically in Fig. 8. This test specimen is similar to those analyzed in the Ref 1 program. Using the CLA routine, excellent correlations have been achieved (Fig. 9) with data from a wide variety of geometries, materials,

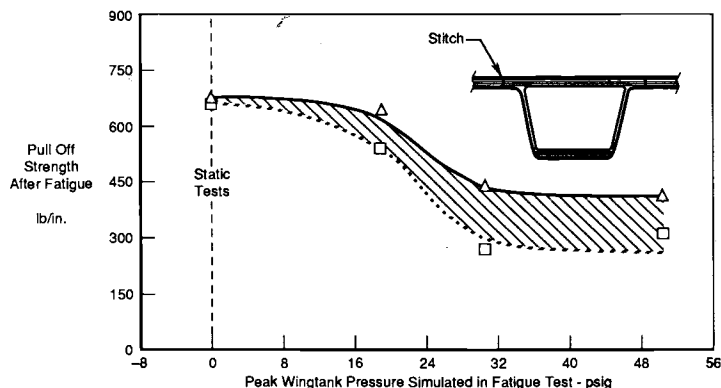


FIG. 6—Effect of fatigue on pull-off strength of hat stiffener panels.

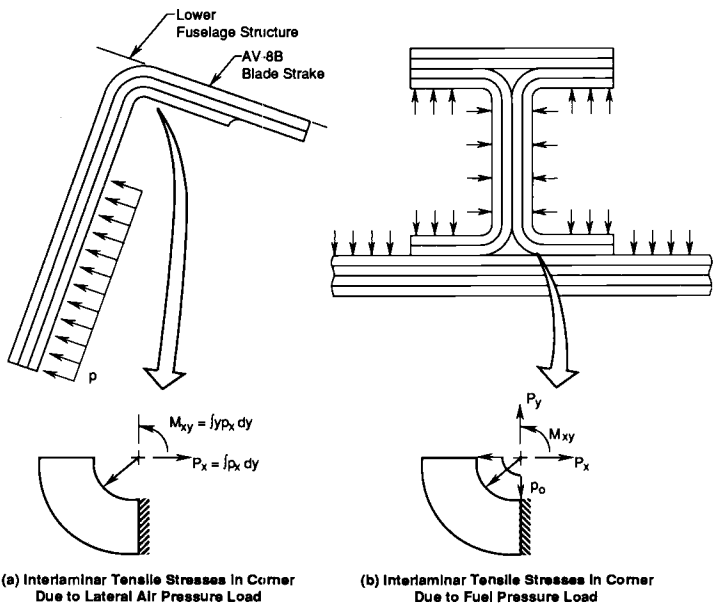


FIG. 7—Curved laminate analysis uses loads defined at the center of the radial arc.

and environmental conditions (Table 1). Specimen dimensions are not critical, as evidenced by the variety of radii noted in the legend to Fig. 9. Typical shear and normal strength properties for AS4/3501-6 are 11 and 3 ksi, respectively, under room temperature/dry conditions. These strengths can be reduced to 70% of these values under elevated temperature/wet conditions.

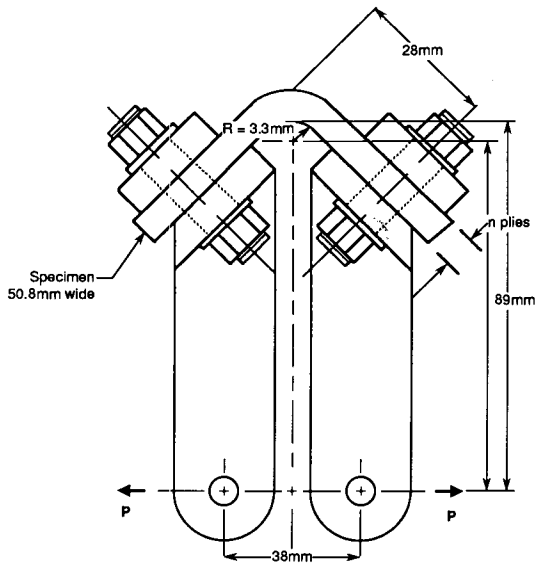


FIG. 8—Schematic of curved laminate test configuration.

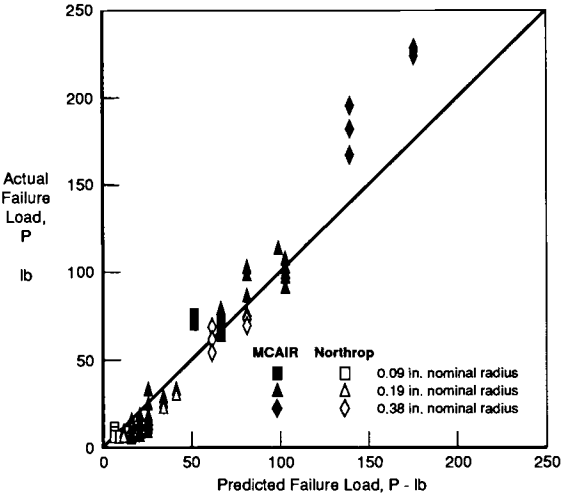


FIG. 9—Curved laminate analysis verification.

TABLE 1—Angle specimen data.^a

Test Case	Layup	r/t^b	Test Condition ^c	P_{actual}, lb^a	$P_{predicted}, lb^a$	f_i
1	20/60/20	1.050	RTW	7.4	7.8	0.95
2	20/60/20	1.453	275/w	7.6	6.2	1.23
3	24/52/24	0.338	RTW	66.0	65.9	1.00
4	24/52/24	0.334	190/w	71.0	52.4	1.36
5	20/60/20	2.675	RTW	11.6	16.1	0.72
6	20/60/20	2.570	190/w	9.4	12.4	0.75
7	20/60/20	2.650	275/w	7.9	9.9	0.80
8	23/54/23	1.892	−65/d	23.8	26.4	0.90
9	23/54/23	1.930	RTW	17.3	26.3	0.66
10	23/54/23	1.969	190/w	14.9	20.8	0.72
11	23/54/23	1.949	275/w	12.1	16.8	0.72
12	21/58/21	0.960	RTW	32.5	41.3	0.79
13	21/58/21	1.160	190/w	27.6	34.8	0.79
14	21/58/21	1.170	275/w	24.6	28.2	0.87
15	24/52/24	0.639	−65/d	94.0	103.0	0.91
16	24/52/24	0.637	RTW	99.0	102.9	0.96
17	24/52/24	0.635	190/w	93.3	81.7	1.14
18	24/52/24	0.637	275/w	71.9	66.8	1.08
19	21/58/21	2.360	RTW	70.1	80.7	0.87
20	21/58/21	2.340	190/w	59.7	62.5	0.95
21	24/52/24	1.248	RTW	225.5	175.8	1.28
22	24/52/24	1.260	190/w	179.3	139.7	1.28

^a 1 lb = 0.45 kg.

^b The value r is inner radius of angle and t is laminate thickness.

^c RTW is room temperature wet XXX°/w is XXX°F/wet.

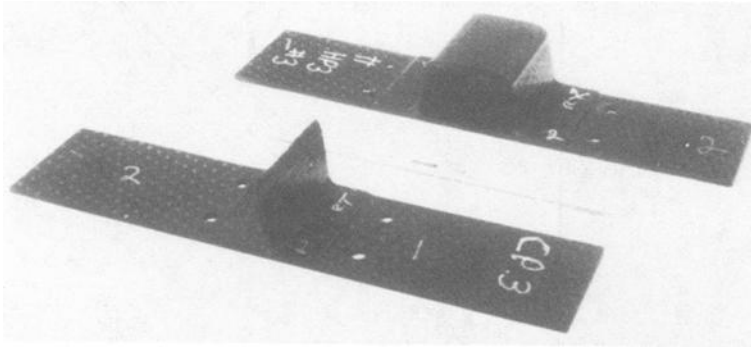


FIG. 10—Co-cured joint element specimens.

The curved laminate test is recommended for aircraft design development programs whenever curved laminates are used as primary load paths. The laminate layup, thickness, and radius at the corner of the specimen should be designed to encompass the range of design applications of the specific structure. The test environment should simulate the service environment of the actual structure. A minimum of five specimens should be tested for each specimen configuration and environment.

Failures in curved laminate radii are not generally catastrophic. The primary failure mode is delamination within the curved section, reducing the bending stiffness of the joint but rarely reducing its axial strength. These delaminations do change the load distributions in the flange attachments. Sometimes, the axial joint strength is enhanced because of the reduced moment. When the primary loading is through shear or an overturning moment, the reduction in strength can result in failure in the curved section.

Stiffener Pull-Off Specimens—Stiffener pull-off specimens (Fig. 10) are designed to verify skin/stiffener separation strength. This specimen and its loading (Fig. 11) are similar to those analyzed during the Ref 1 program.

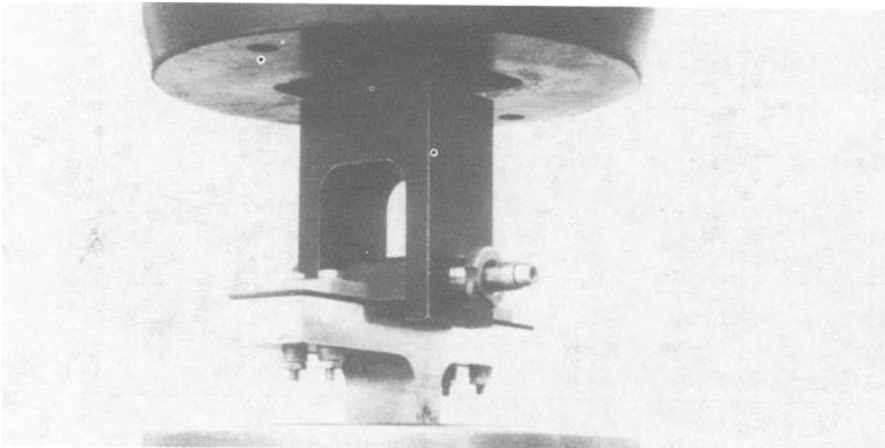


FIG. 11—Direct out-of-plane loading of hat-stiffened element specimen.

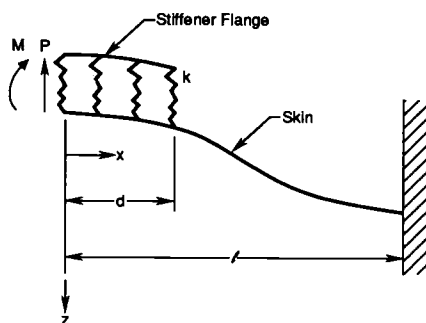


FIG. 12—Interfacial stress field model (ISFM) for direct load cases.

FEA showed that the direct hat pull-off test induces a moment beneath the web of the hat that increases stresses by as much as a factor of three above that computed without considering this moment. The stress distribution created by this moment is not predicted with simpler methods, such as ISFM (Fig. 12). The SSAM routine was developed specifically to examine that stress distribution. While the simpler, closed-form analysis methods can be adjusted to provide reasonable estimates of the failure loads for skin/stiffener separation (Fig. 13), we are not certain that the stress distribution created beneath the web of the hat in the pull-off test is representative of the distribution created beneath the stiffener of a buckled or laterally loaded panel.

Because of the difference in skin/stiffener joint loading between a direct pull-off test and the pull-off provided by postbuckled panels, a different loading condition (Fig. 14) is recommended for future tests. The pull-off strength is determined using a failure mode induced by the buckle shape of the skin, rather than direct pull-off loads, as shown in Fig. 10. The length and loading conditions are selected to match the buckle shape of the skin as closely as possible. The purpose of this test specimen configuration and loading is to simulate the loading experienced in actual postbuckled panels.

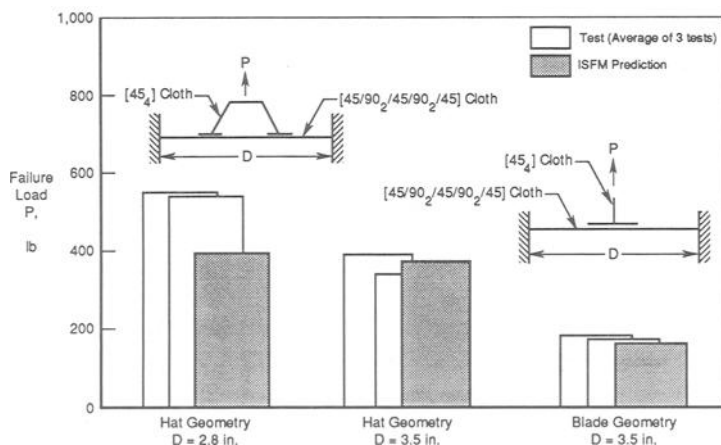


FIG. 13—Correlation of predicted and measured pull-off strengths.

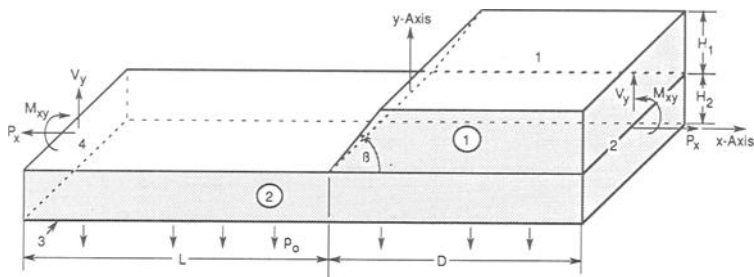


FIG. 16—Local elasticity model for stiffener termination analysis.

strength of the stiffener at any particular transverse section. This analysis uses the out-of-plane load distribution determined from the longitudinal analysis. The stiffener runout strength is determined by applying a Tsai-Hill-type quadratic failure criterion to the normal and shear stress distributions obtained from the transverse analysis

$$\sigma_N = F_N \sqrt{1 - \sigma_S/F_S)^2}$$

where

- σ_N and F_N = the allowable normal stress and measured normal strength of the matrix material and
- σ_S and F_S = the applied shear stress and measured shear strength of the matrix material.

Figure 17 shows the results of the correlations between this method and stiffener runout test data.

Tests of stiffener runout specimens (a simplified version of which is recommended and shown schematically in Fig. 18) are used to verify the runout design and the analytical predictions. These tests should include a sufficient range of termination angles to optimize the structural design.

Postbuckled Panel Subcomponents—Out-of-plane loads can arise indirectly and cause failure. The out-of-plane deformations of buckled panels induce loads on stiffeners and their

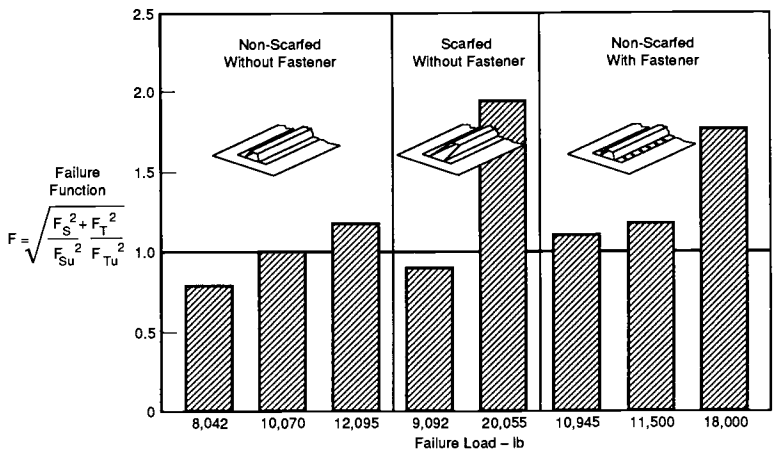


FIG. 17—Compression bar stiffener runout specimen failure correlation.

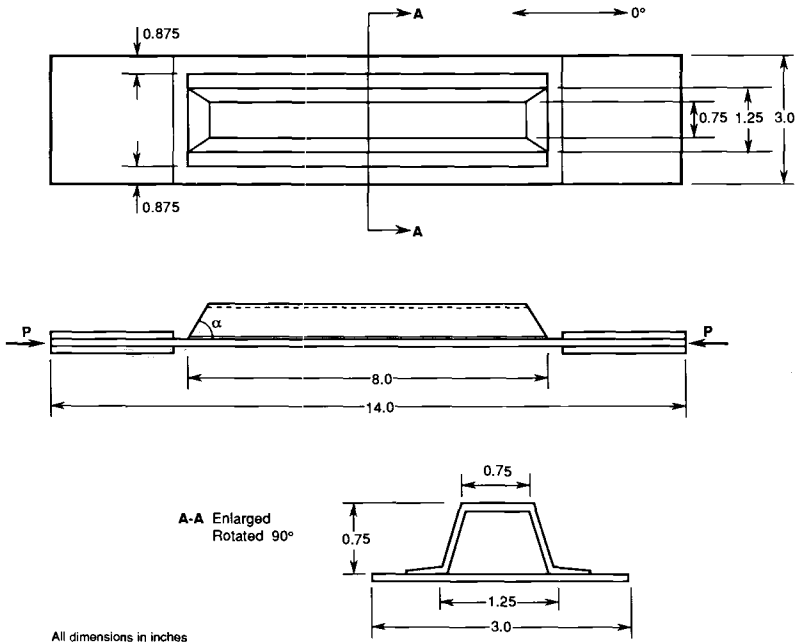


FIG. 18—Schematic of stiffener runout test specimen.

attachments. These loads can lead to separation of the skin and stiffener, and ultimate failure. The ISFM method was modified to analyze these failures. Figure 19 summarizes the model geometry and boundary conditions assumed within ISFM for this case. This figure also shows how the integrally stiffened buckled panel is modeled for analysis by ISFM.

Results of combined loading postbuckling tests of flat multibay composite panels were used to correlate this analysis (Fig. 20). Because experimental data for the interfacial shear modulus,

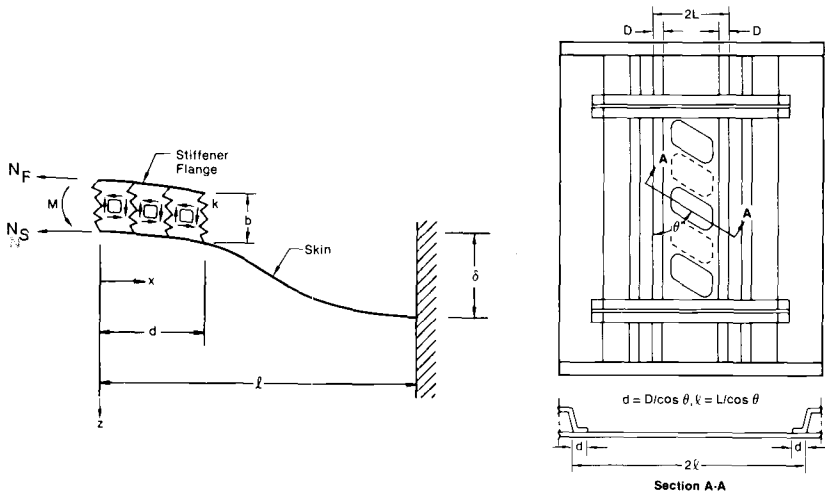


FIG. 19—Interfacial stress field model for indirect load cases and its application to a buckled panel.

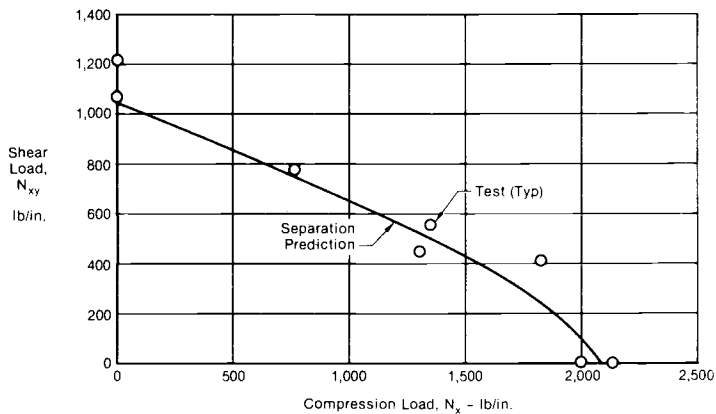


FIG. 20—Correlation of ISFM and test data of panels subjected to combined loads.

elastic foundation modulus, and interface thickness were not available, values were assumed such that the analysis matched the results of the pure compressive and the pure shear load cases. The resulting analysis correlated well with the combined load cases. The modulus assumptions could have been avoided had experimental data been available.

Conclusions

The methods and experience gained in this program indicate the need for new test techniques to develop the mechanical property data required to characterize stresses and displacements in co-cured structures. Whereas a number of techniques have been standardized for bonded joints, no such database exists for co-cured structures or for laminates subjected to out-of-plane loadings. And while a number of element test techniques examined in this program gave reasonable strength data, no stiffness data were obtained. Because local out-of-plane stresses are sensitive to out-of-plane stiffness (or modulus), experimental determination of the moduli using simple coupon specimens is crucial.

The authors recommend that ASTM pursue standard test methods required to develop confidence in the strengths and mechanical properties of co-cured composite joints. The test procedures, data, experience, and allowables development techniques (even for current materials) would pay large dividends in the design of lightweight, durable, and reliable aircraft structures.

Acknowledgment

The information presented herein was generated under a joint program sponsored by the Naval Air Development Center and the Federal Aviation Administration.

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Analysis and Testing of a Composite Sandwich Shell Horizontal Tail

REFERENCE: Sherrouse, M. A. and McWhorter, J. C., III, "Analysis and Testing of a Composite Sandwich Shell Horizontal Tail," *Composite Materials: Testing and Design (Tenth Volume)*, ASTM STP 1120, Glenn C. Grimes, Ed., American Society for Testing and Materials, Philadelphia, 1992, pp. 253–263

ABSTRACT. This paper presents the preliminary results of an investigation into the use of a sandwich shell without spars or ribs as a horizontal tail on a light general aviation aircraft. The primary motivations for designing such a structure are simplicity, ease of construction, and the reduction of fasteners.

The structure investigated consisted of two symmetric, quasi-isotropic sandwich skins with four plies of graphite-epoxy unidirectional prepreg per face and 10-mm honeycomb core laid up in an airfoil-shaped mold and cured at 120°C. The two halves were joined by adhesive to form a tapered cantilevered beam with airfoil cross section truncated at the 56.7% chord line. Two internal ribs were installed at the fuselage contour near the center of the tail to provide attachment to a frame. Symmetric point loads were applied to produce varying amounts of torsion and bending, and the results were compared with analysis from a NASTRAN model using elements with composite material properties. The structure was loaded to failure, and the results were compared with analysis from a NASTRAN buckling routine.

A structurally intact portion of the failed test piece was modified to create one half of a full-size forward-swept tail. Again, experimental results from various bending and torsion loads were compared with analysis from a comparable NASTRAN model.

Chordwise and spanwise strain plots indicated that the sandwich skins carry the bending loads, whereas the reinforced areas at the leading and trailing edges act as shear webs. Good correlation between the experimental strains and NASTRAN analysis was obtained.

KEYWORDS: composite materials, sandwich skin, shell structure, aircraft horizontal tail, finite-element analysis (NASTRAN), forward sweep

The aerodynamic surfaces of an aircraft (that is, the horizontal and vertical stabilizers and wings) are some of the most difficult components to manufacture. Standard design procedure calls for numerous ribs and at least one spar as internal stiffeners, each of which has to be constructed to precise dimensions. The higher specific stiffness of graphite-epoxy composite materials introduces the possibility of using sandwich shell structures for some aerodynamic surfaces. Some modern supersonic fighter aircraft already incorporate tail sections that have full-depth honeycomb between the upper and lower skins (see Ref. 1). This is made practical because of their exceptionally thin cross section and high aerodynamic loading. For thicker aerodynamic surfaces, however, it may be more efficient to use sandwich skins and hollow interiors. By eliminating the spars and ribs, production is dramatically simplified, structural weight is reduced, and far fewer fasteners are required. Also, because stiffness is continuous throughout most of the structure, local stress concentrations caused by the interaction between

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the skin and stiffeners, including ribs and spars, are significantly reduced (see Ref 2). Problems such as the economics of successfully fabricating large composite shells, moisture absorption by the sandwich core, damage tolerance, and repairability have to be considered before sandwich shell structures replace traditional skin-stringer designs.

This paper presents preliminary results of an investigation into the use of an all-composite, sandwich shell without spars or ribs as a horizontal tail on a light general aviation aircraft. Various point loads were applied to a full-span tail, and the experimental results were compared with analysis from an MSC/NASTRAN finite-element model. Buckling strength, adhesive joint strength, and deformation characteristics were three areas of particular interest to the investigators. The effect of forward sweep on this type of structure was also studied. The experiments performed were divided into three basic groups, hereafter to be referred to as Tests 1, 2 and 3. These are described in more detail in the next section.

No attempt was made to apply aerodynamic loads to any of the models or test specimens. Nor is any inference intended as to the dubious benefits that may be obtained through the use of a forward-swept horizontal stabilizer. Except for the shape of the upswept tail, which is modeled after the tail of an existing aircraft, all aspects of these tests were kept as generic as possible. Also, these preliminary studies were intended to locate possible stress concentrations and other problem areas, not to describe them accurately. That has been left to future research.

Procedure

The external dimensions of the test piece used in the first two tests were modeled after the horizontal stabilizer of a typical light general aviation aircraft, without the moveable control surfaces. Figure 1 shows the test piece divided into its individual components. Although the tail appears to have sweep due to the absence of the control surfaces and the angle of the fuselage intersection, it is in actuality unswept because the mean aerodynamic chord line is perpendicular to the airflow across the tail. The widths of the structure at the center span and tip were 62 and 45 cm (24.375 and 17.75 in.), respectively. The distance from center span to tip was 170 cm (67 in.). The cross section of the assembled structure was that of a NACA 0009 airfoil (a symmetric airfoil with 0.09 thickness-to-chord ratio) truncated at the 56.7% chord line. The maximum thickness of the airfoil varied from 9.82 cm (3.87 in.) at center span to 7.15 cm (2.82 in.) at the tips. Throughout this paper, three reference axes are referred to: span-wise, which is measured from mid-span along a line of constant percent chord; chordwise,

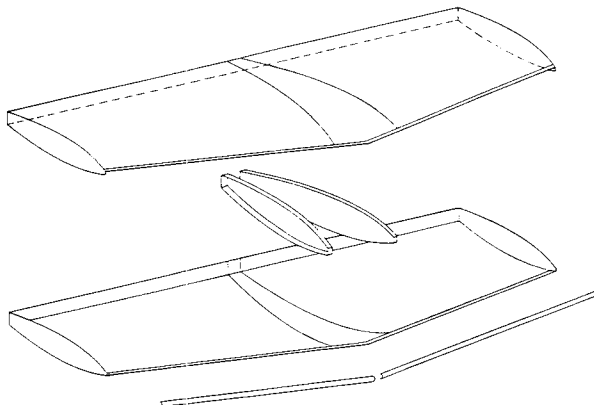


FIG. 1—Exploded view of full-span test piece.

which is measured from the leading edge along a line parallel to the center span line, and transverse, which is perpendicular to the other two axes

The top and bottom skins consisted of quasi-isotropic graphite facings on a 10-mm ($\frac{3}{8}$ -in.) Hexcel HRH-10-1/8-3 0 hexagonal cell Nomex honeycomb core. Each facing was 0.66 mm (0.026 in.) thick and consisted of four plies of Fiberite HY-E 30714A unidirectional graphite-epoxy prepreg. These were laid up in an airfoil-shaped mold and cured at 120°C (250°F). Because of cost constraints, the only material available for this project was graphite epoxy in excess of one year old that had been stored at -17.8°C (0°F). The finished shell had some dry spots with minor defects, which were judged through visual inspection to be unimportant for the purpose of observing the large-scale behavior of the test piece. No tension test specimens of the material were laid up at the time of construction, although results from tension tests performed when the material was originally purchased were available [3]. After the test piece was broken, several tension specimens were cut from the failed test piece to approximate the actual material properties.

The top and bottom halves were joined by Dexter Hysol Epon 901/B-1 general-purpose structural adhesive at the overlapping sections of the rear spar and tips, and in the leading edge cap. Two internal ribs were used at the fuselage contour near the center of the tail to enable attachment to a frame that would simulate the actual juncture with a fuselage. The test piece was bolted through the flange of the rib to a 3-mm ($\frac{1}{8}$ -in.) thick right-angle bracket made from bidirectional graphite-epoxy cloth. This bracket was bolted to a frame constructed from 10-mm ($\frac{3}{8}$ -in.) steel plates welded together with the load-bearing sides at a 97.5° angle to the center line of the span.

Unidirectional strain gages aligned in the longitudinal, or spanwise, direction were applied to the internal and external surfaces of the top, right-hand sandwich skin. Thirty-six 120- Ω strain gages were attached on the internal and external surfaces in a spanwise row along the 25% chord line, and in a chordwise row at the 25-cm (10-in.) span location. Seven additional gages were also installed in a chordwise row at the 75-cm (30-in.) span location on the external surface. The leads to the gages on the internal surface were run out of the tail through a 10-mm ($\frac{3}{8}$ -in.) diameter hole in the center of the rear spar at the 50-cm (20-in.) span location.

Test one consisted of a series of point loads that produced varying amounts of bending and torsion on the structure. The strain gages were attached to a data logger, which passed the acquired strain data on to a portable personal computer (PC). The loads were applied by placing lead bars of known weights in a sling hanging from a wooden bracket. The purpose of the bracket was to distribute the applied load over a suitable area to minimize local stress concentrations. All loads were applied symmetrically on either side of the support frame. The first experiments were done with the bracket at the 120-cm (47.5-in.) span location. Three series of strain measurements were taken with the loads applied at the leading edge, trailing edge, and approximate shear center. The bracket was then moved to the tip, and another series of measurements was made with the load at the approximate shear center.

A half-span NASTRAN model of the test piece was created for comparison. The model was generated on a PC using MSC/MOD, a preprocessor for MSC/PAL2 with the capability of translating to a NASTRAN input deck. Only the right half of the span was modeled because the applied loads and boundary conditions were symmetric about the center line of the span. The boundary conditions for all nodes at center span involved fixing translation in the spanwise direction and rotation about the transverse and chordwise axes. Also, translation in the chordwise and transverse directions, and rotation about the spanwise axis were zeroed at the rib. To simplify the implementation of the boundary conditions, the spanwise direction was considered to be perpendicular to the chordwise direction. Quadrilateral plate elements were used almost exclusively to avoid problems with material axis orientation associated with triangular plate elements. Coordinates for the airfoil cross section were obtained using the equa-

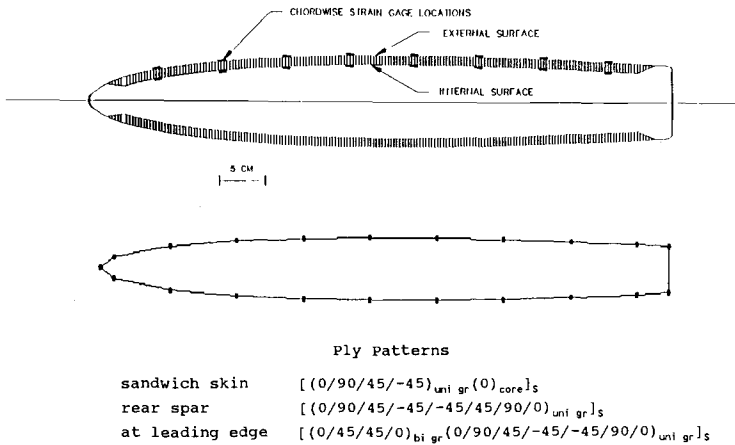


FIG. 2—Comparison of cross section of test piece and finite-element model.

tion for a NACA 0009 airfoil found in Ref 4. All spanwise rows of nodes were positioned along constant percent chord lines.

After translation to NASTRAN coding, the deck was modified slightly to incorporate PCOMP composite laminate cards and MAT2 anisotropic material property cards. The PCOMP cards of all elements adjacent to the leading edge contained the ply pattern of the leading edge cap as well as the ply pattern of the skins without the honeycomb core. All rear spar and tip element PCOMP cards consisted of twice the ply pattern of the skins without the honeycomb core. The varying properties of the transition region of the skins, where the honeycomb is reduced to zero thickness, were not modeled, nor was the small region near the trailing edge where there was no honeycomb (Fig. 2). It was necessary to alter the material axis of each spanwise row of elements slightly to keep the material properties' reference axes parallel over all of the surfaces. The model was analyzed using MSC/NASTRAN Version 65-C, and the appropriate data were plotted.

The second test was performed with the tail section mounted in a structural test stand. The tail support frame was welded to the center of the test frame. The load was applied to the center of both tips through an aluminum I-beam whiffletree loaded at its center by a hydraulic jack. A load cell was installed at the right-hand tip to monitor the applied load per tip. Twenty-two more spanwise 120-Ω strain gages were attached to the external bottom surfaces of both the left and right halves of the test piece along the 25% chord line. Also, seven more chordwise gages were added to the external bottom skin on the right side of the tail at the 25-cm (10-in.) span location. Acoustic sensors were also attached to the external surfaces of the test piece to record the intensity of acoustic emissions. In this manner, it was hoped to determine the rate of fiber failure occurring during the test. However, this was done primarily to gain experience with the equipment and techniques involved in acoustic monitoring.² The test piece was loaded in approximately 50-kg (110-lbs) increments from 0 kg to failure, with specified holds and unloads for the analysis of acoustic emissions.

The NASTRAN model from the first test was also used for the second test, with the exception of a few modifications in the control cards. The control cards were changed to facilitate the use of the NASTRAN linear buckling routine. The SINV eigenvalue extraction technique, which uses the Sturm inverse method, was used to determine the buckling strength of the test piece.

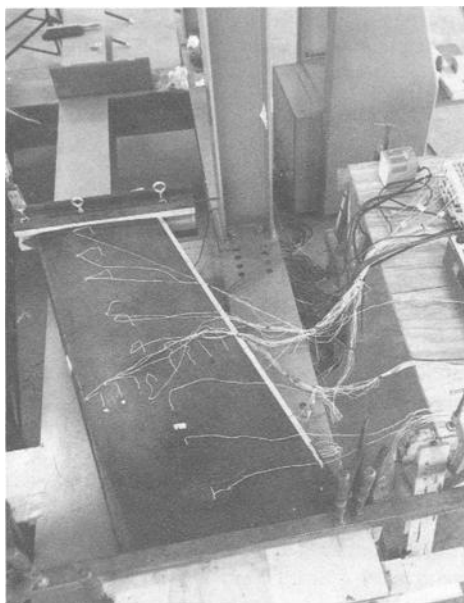


FIG. 3—Forward-swept tail mounted in test frame.

After the test to failure of the tail, the structurally intact right side was modified to create a half-span forward-swept tail. A wooden block, contoured to match the internal shape of the airfoil shell, was inserted into the root section of the test piece and bonded in place. An external wooden support frame was contoured to the external shape and bonded in place to provide as rigid a support as possible. The support block was set such that the tail was at 10° of forward sweep. The complete structure was firmly attached to the base of the test frame using numerous C-clamps (Fig. 3). The chordwise strain gage nearest the leading edge on the top external surface at the 25-cm (10-in.) span location was removed during the alteration process, but all other gages remained intact or were repaired before the tests resumed.

The NASTRAN model used in the previous tests were modified for the forward-swept tail by truncating along the clamped edge, and altering the nodes and elements as necessary. Figure 4 shows the nature and extent of the modifications to the geometry of the model. The bound-

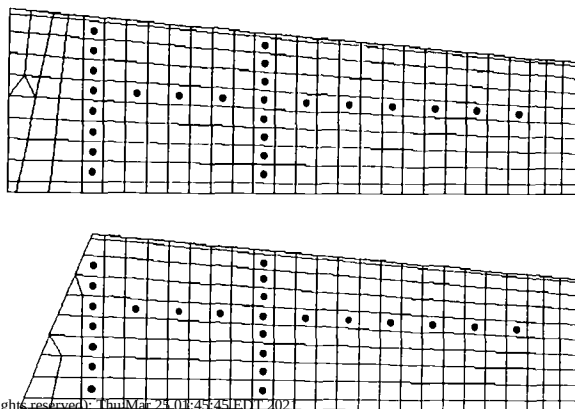


FIG. 4—Planform of finite-element models for unswept and forward-swept tails.

ary conditions were also changed such that all degrees of freedom along the clamped edge were set to zero. Although this did not precisely model the flexibility of the support frame and C-clamps, it was deemed satisfactory for the scope of these tests.

Analysis of Results

Except where noted, all strains plotted in the accompanying graphs are the absolute value of the strains divided by the load. For those gage readings that became nonlinear, only the linear portion of the data was considered. The results from test two and the tip loading of test one are plotted together because there were only slight differences in their loading and boundary conditions. Note, however, that in Test 1 the top surface was in tension and the bottom surface was in compression; for Test 2 the opposite is true. Also, because the NASTRAN results from both tests are identical, they are only plotted as one curve.

The effect of load position on the chordwise distribution of longitudinal strains can be seen in Figs. 5 and 6. As a result of the warping restraint effect of the support structure, the sensitivity to torque decreases significantly between the 25- and 75-cm (10- and 30-in.) span locations. In Fig. 7, the NASTRAN values using the different material properties for the unidirectional graphite-epoxy are compared. All subsequent plots display only the NASTRAN values using the revised material properties (that is, the material properties experimentally determined from test specimens taken from the failed test piece).

The internal and external chordwise distributions of longitudinal strains at the 25-cm (10-in.) location for Tests 1 and 2 are compared in Fig. 7. There is little change in the results between the two sets of experiments despite the switch from downloading to uploading of the test piece. The offset between the NASTRAN results and the experimental results is most likely due to the inaccuracy of the material properties determination. The NASTRAN results do, however, reflect the same general trends as the experimental data. This is also true for the spanwise strain distributions for the same surfaces (Fig. 8).

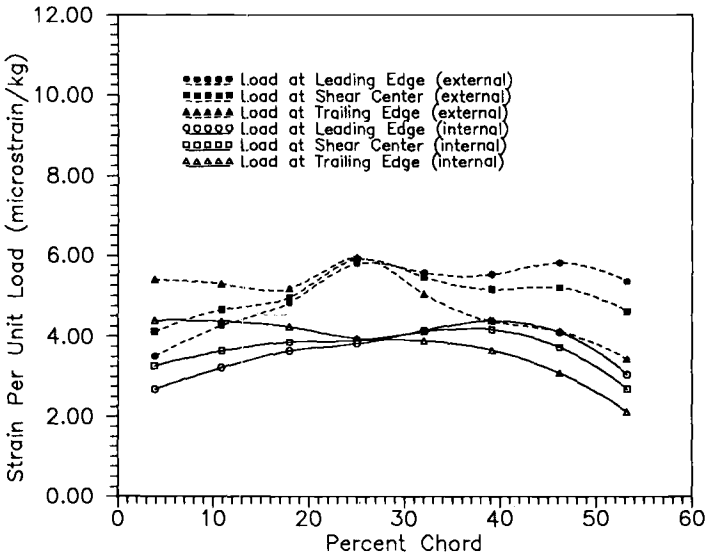


FIG. 5—Test 1 experimental chordwise distribution of longitudinal strains at the 25-cm (10-in.) span location

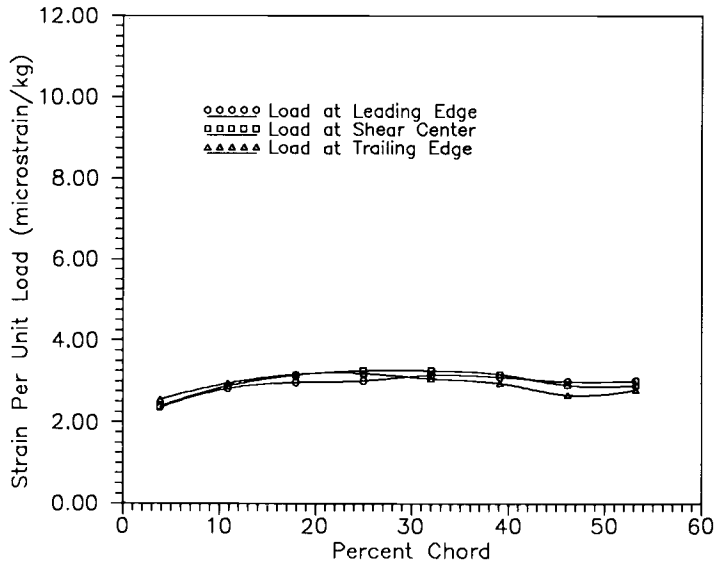


FIG. 6—Test 1 experimental chordwise distribution of longitudinal strains at the 75-cm (30-in.) span location.

Strain per unit load versus the applied load on each tip is plotted in Fig. 9 for several gages located in what appeared to be a buckled region. Nonlinear behavior of the tail is clearly indicated by this figure. During the testing, there was a visible flattening of the upper surface, which gradually increased with load rather than a sudden buckle. Because this deformation could alter the aerodynamic characteristics of the tail, interesting ramifications of this phenomenon

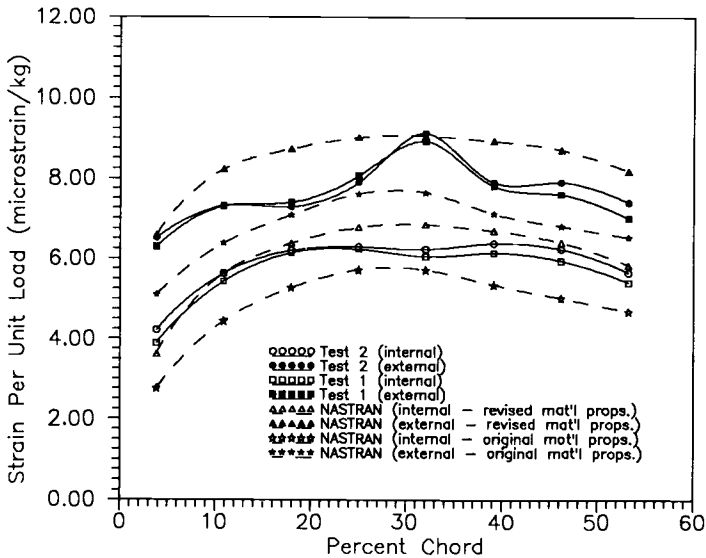


FIG. 7—Comparison of Tests 1 and 2 chordwise distribution of longitudinal strains at the 25-cm (10-in.) span location.

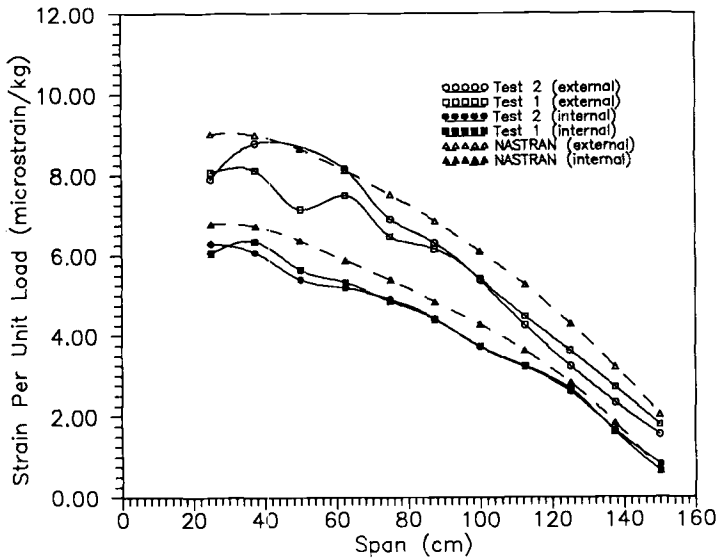


FIG. 8—Comparison of Tests 1 and 2 spanwise distribution of longitudinal strains.

during flight conditions are possible. The failure of the test piece was probably caused by high plate bending stresses and ply failure at locations with local defects. Playback of videotape of the moment of failure indicated that failure began as two widely separated cracks and progressed rapidly at -45 and 45° angles toward the mid-chord of the tail. The failed area of the test piece can be seen in Fig. 10. The maximum longitudinal strains recorded for these experiments were less than 3500 microstrain at locations well removed from any strain concentration. It is likely that much higher strains were present near the plate-web intersection. Also, data from the acoustic emissions tests seemed to indicate that excessive ply fracture had not

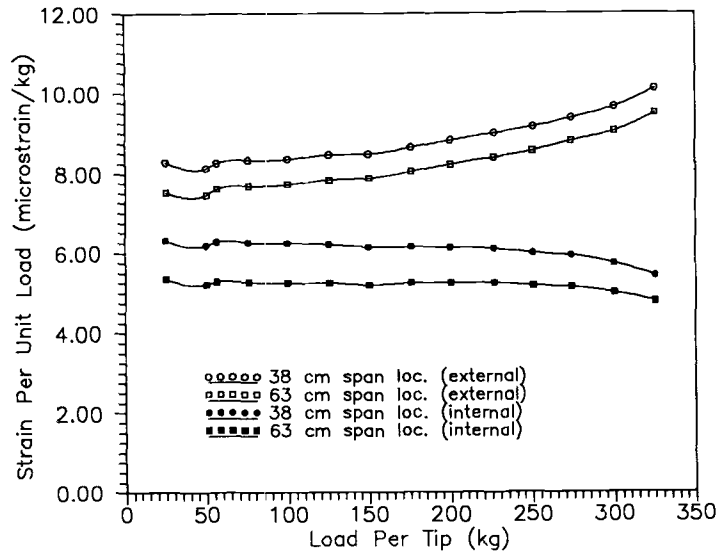


FIG. 9—Strain per unit load versus applied load per tip.



FIG. 10—Full-span test piece after failure.

occurred. This evidence implies that the failure of the structure was of a local nature and somewhat premature when compared with its potential strength. Failure occurred with a load of approximately 365 kg (800 lbs) on each tip, which is below the NASTRAN predicted value of 545 kg (1200 lbs). The magnitude of this discrepancy may be attributable to the use of the NASTRAN differential stiffness buckling routine. A postbuckling or nonlinear analysis would probably yield more accurate results.

The results from the forward-swept tests demonstrate the effect of the change in boundary conditions. Figures 11 and 12 compare the longitudinal strains for the internal and external chordwise rows of gages at the 25-cm (10-in.) span location for the forward-swept and unswept cases. The most noticeable effect is on the external surface near the leading edge. This effect

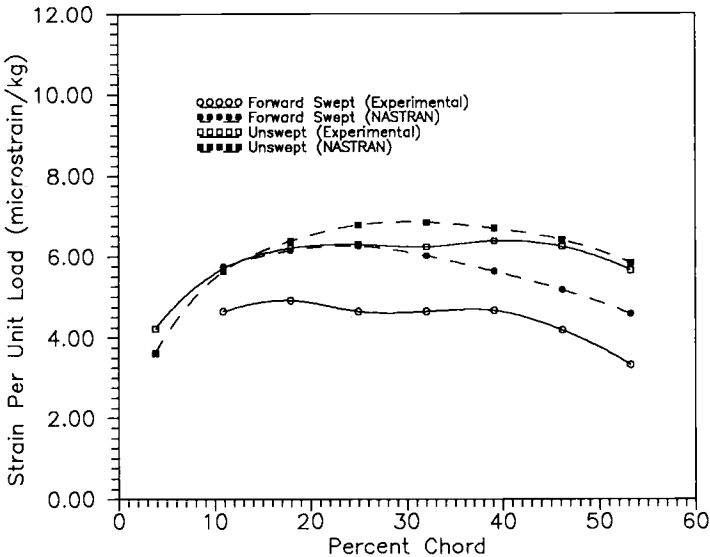


FIG. 11—Comparison of Tests 2 and 3 chordwise distribution of longitudinal strains on the internal surface at the 25-cm (10-in.) span location.

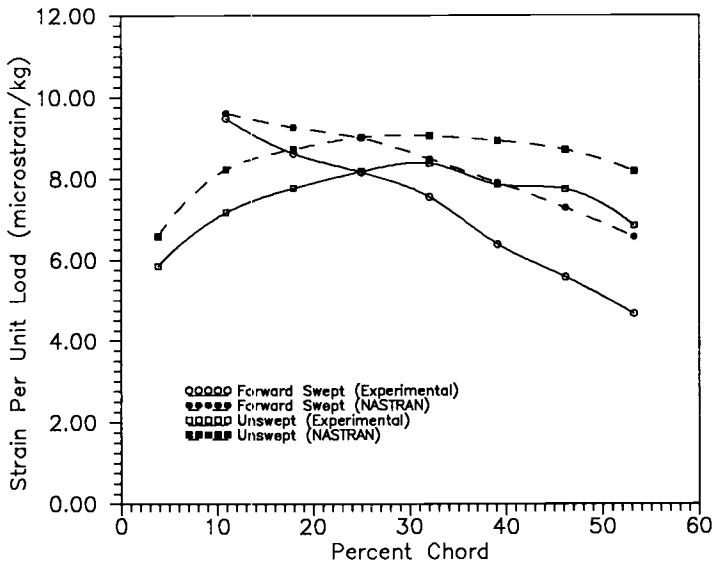


FIG. 12—Comparison of Tests 2 and 3 chordwise distribution of longitudinal strains on the external surface at the 25-cm (10-in.) span location.

decreases by the 75-cm (30-in.) span location (Fig. 13). Note that the support on the forward-swept tail is at a different relative distance from the gages than that on the unswept tail. Figure 14 shows the effect of chordwise travel of the applied load on longitudinal strains at the 25-cm (10-in.) span location [5]. The highest relative strains were obtained with the load at the leading edge. This is an important consideration because aerodynamic loads are usually much higher near the leading edge.

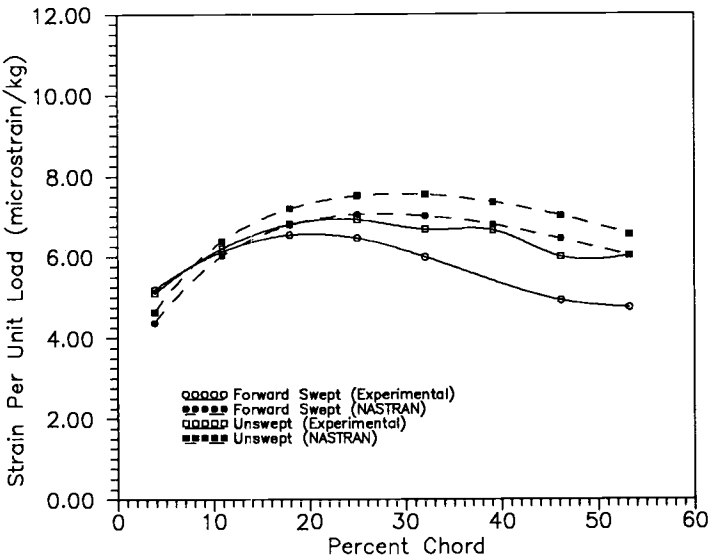


FIG. 13—Comparison of Tests 2 and 3 chordwise distribution of longitudinal strains on the external surface at the 75-cm (30-in.) span location.

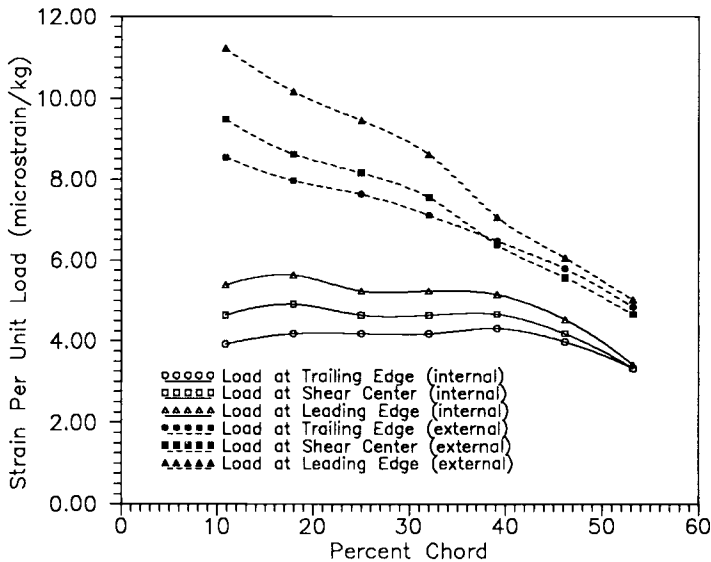


FIG. 14—Experimental chordwise distribution of longitudinal strains at the 25-cm (10-in.) span location on the forward-swept test piece.

Conclusions

The goal of simplifying construction was definitely achieved. Also, despite the failure to obtain accurate material properties at the time of construction, the agreement between NAS-TRAN and the experimental results was generally good regarding trends in the data. It is anticipated that with carefully determined material properties, more accurate results will be attained in the continuation of this research. The adhesive joints did not fail during the tests; however, more detailed studies need to be done of these and other possible failure areas. It was determined that the rear spar and the reinforced area of the leading edge acted as shear webs, whereas the bending loads were carried by the sandwich skins. Further study of the areas immediately adjacent to the honeycomb and shear web is recommended.

It is imperative to the verification of feasibility of this concept that further studies be made into the extent and consequences of cross-sectional variation under high aerodynamic loads. One cure for this problem would be to increase the thickness of the honeycomb core in the sandwich skins. Although many areas need to be investigated, the potential advantages of monocoque construction should stimulate continued interest.

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Composite Material Stub-Blade Wing Joint

REFERENCE: Franklin, W. M. and Kreimendahl, B. W., "Composite Material Stub-Blade Wing Joint," *Composite Materials Testing and Design (Tenth Volume)*, ASTM STP 1120, Glenn C. Grimes, Ed., American Society for Testing and Materials, Philadelphia, 1992, pp. 264–278

ABSTRACT: A design study to determine the viability of a stub-blade wing joint for HALE (high-altitude, long-endurance) aircraft is presented. The resulting wing joint design uses advanced composite materials and is capable of carrying moderately high bending, shear, and torsional loads. The stub-blade concept is similar to the wing joint design used on sailplanes and is particularly applicable to a large-span, high-aspect-ratio, unswept wing that uses a single-cell wing box to carry spanwise loads. This wing structure and geometry are typical of many HALE aircraft. The stub-blade joint can be located at any wing station and uses "pin-in-a-socket" action to transfer loads from the outboard to inboard wing box. Justification and criteria for material selection are presented. Extensive use of graphite/epoxy, co-cured and bonded structure is made to achieve a minimum weight design. Major design drivers include minimum weight, ease of fabrication, and ease of repeated assembly and disassembly of the joint. The latter would be desirable on a large-span aircraft for which hangar or transportation limitations would require repeated removal of outer wing sections. A FORTRAN program was developed to optimize the blade length based on weight, and a NASA Structural Analysis (NASTRAN) model of the stub-blade design was constructed to investigate the distribution of internal loads. The stub-blade joint presented in this paper was sized to loads generated for a representative baseline HALE vehicle. This Lockheed Baseline HALE Aircraft is an unmanned vehicle capable of reaching an altitude of 27,432 m (90,000 ft) and features a 8163-kg (18,000-lbs) gross weight and 80.5-m (264-ft) wingspan. The stub-blade wing joint design is weight competitive with the traditional tension-type wing joint and offers attractive features such as ease of assembly and disassembly, ease of fabrication, and aerodynamic smoothness.

KEYWORDS: composite materials, aircraft, wing joints, design, HALE aircraft, optimization

Background

The design of wing joints has evolved into two primary types: the tension joint (Fig. 1, used on aircraft such as the Lockheed F-104, U-2, and General Dynamics F-16) and the shear joint (Fig. 2, used on aircraft such as the McDonnell Douglas F-15 and F-18). Some aircraft (Lockheed L-188 Electra and L-1011 TriStar) use a combination of these two types in their wing joint design. The tension joint requires accurate mating surfaces for fastener preload and uses numerous attachments. The shear joint requires extremely accurate alignment of lugs. Both the tension and shear joints are considered to be semipermanent because the number of attachments and required close alignment of mating surfaces makes repeated assembly and disassembly very difficult.

Lockheed Aeronautical Systems Company envisions that large-span aircraft of the future,

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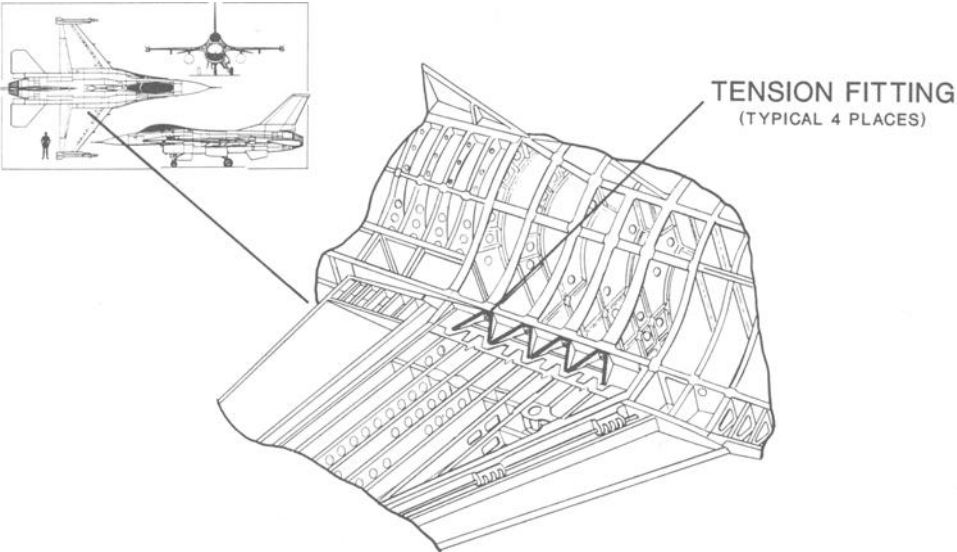


FIG. 1—Traditional tensile-type wing joint (General Dynamics F-16).

such as manned or unmanned HALE (high-altitude, long-endurance) aircraft, will require repeated removal and replacement of outer wing sections because of hangar and transportation limitations. Proposed HALE aircraft designs with wingspans of 61 m (200 ft) or greater are not uncommon. True high-altitude, long-endurance performance will also require a vehicle with an extremely low structural weight fraction, and this low weight fraction will result in

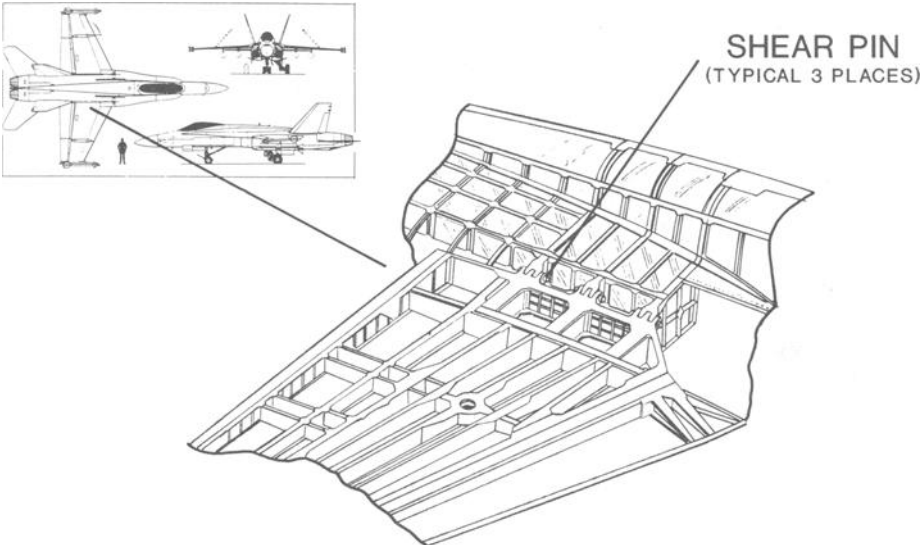


FIG. 2—Traditional shear-type wing joint (McDonnell Douglas F-18).

a relatively fragile aircraft. Hangars or protective enclosures are necessary to protect the structure from the possible damaging effects of wind, rain, and snow and also to minimize exposure to ultraviolet (UV) radiation and temperature variations. In addition, deployment of these aircraft may involve ground or air transportation, also requiring removal of outer wing sections.

Many candidate HALE aircraft missions, such as reconnaissance and surveillance or communications relay, may require that the aircraft be ready for flight with minimal delay after being removed from the hangar or transportation enclosure. Therefore, an alternative wing joint design that offers improved ease of assembly and disassembly compared to the traditional shear and tension wing joints must be developed. In addition, HALE aircraft will require extensive use of advanced composite materials to meet aggressive weight and performance objectives. Consequently, the alternative wing joint design should lend itself to fabrication using these materials.

Stub-Blade Wing Joint Design Concept

The stub-blade wing joint is similar in concept to the wing joint design used on modern sailplanes. This joint uses “pin-in-a-socket” action to transfer loads from an outboard wing section to an inboard wing section (Fig. 3). This joint design is especially applicable to the large-span, high-aspect-ratio, unswept wing design proposed for many HALE aircraft. In addition, the candidate wing would use a single-cell wing box to carry spanwise loads. Assembly and disassembly of the stub-blade joint are possible with a single pin. This pin eliminates the difficulty in mating wing sections that traditional tension and shear joints exhibit. A free-body diagram showing how the stub-blade joint reacts loads corresponding to all six degrees of freedom ($M_x, M_y, M_z, P_x, P_y, P_z$) is shown in Fig. 4.

Lockheed Aeronautical Systems Company has calculated that a structural wing weight of

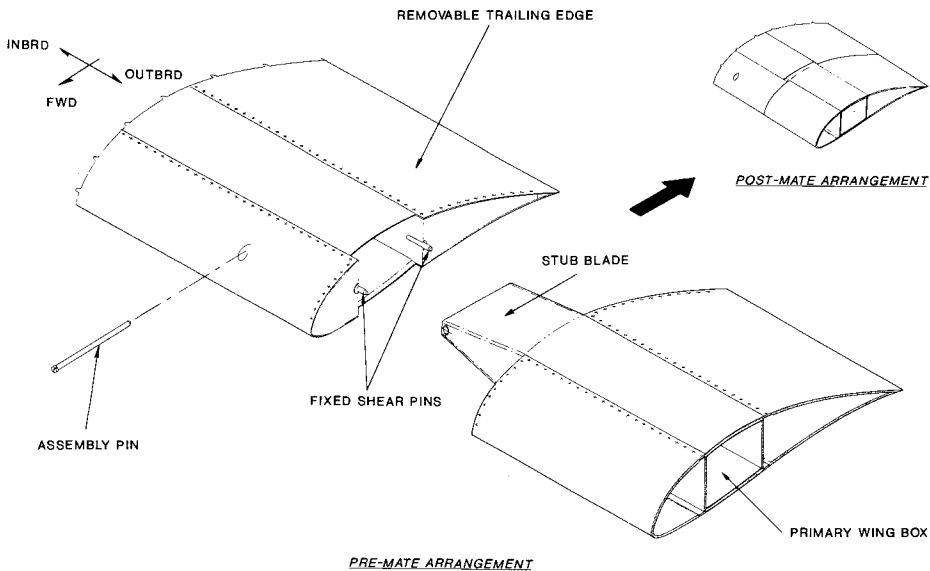


FIG. 3—Stub-blade wing joint structural concept.

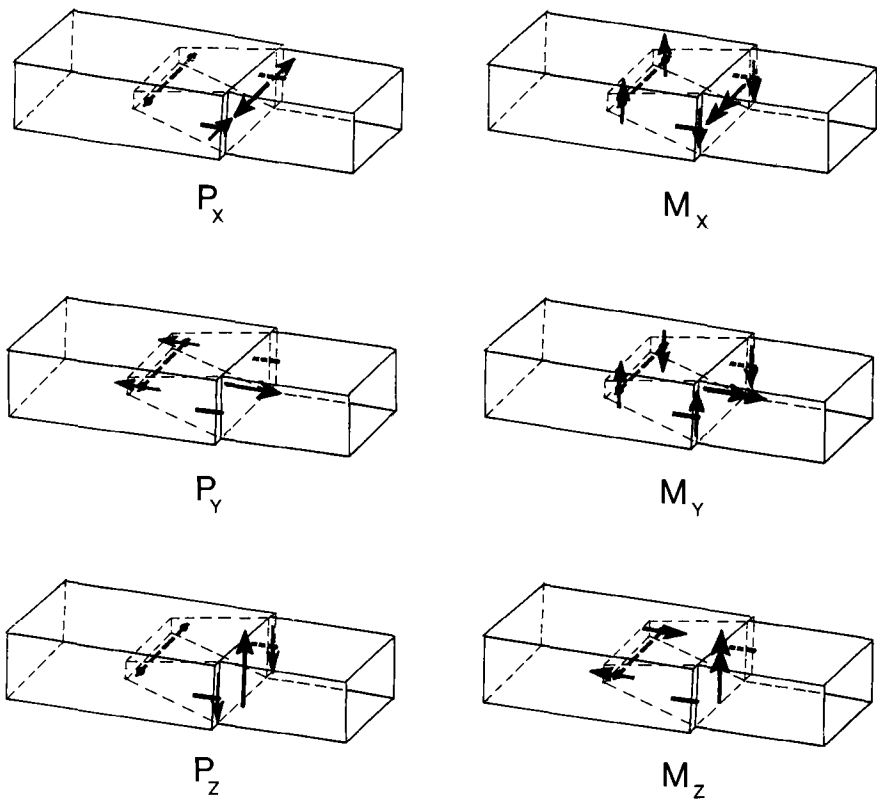


FIG. 4—Freebody diagram of stub-blade wing joint concept.

not more than 7.3 to 9.8 kg/m² (1.5 to 2.0 lb/ft²) is necessary for an aircraft to achieve an altitude of 27 432 m (90 000 ft). Extensive use of advanced composite materials is necessary to attain this low wing weight. The stub-blade joint concept represents a practical composite material design because it is an easily manufactured co-cured and bonded structure requiring a minimum of mechanical fasteners.

Baseline Aircraft Geometry

The objective of this design study was to determine the viability of the stub-blade wing joint concept based on design loads and wing construction of a representative baseline HALE aircraft. To this end, the Lockheed Baseline HALE Aircraft was defined (Fig. 5). The Lockheed Baseline HALE Aircraft is an unmanned vehicle that features a wing span of 80.5 m (264 ft) and a gross vehicle weight of 8165 kg (18 000 lbs), and is predicted to be capable of reaching an altitude of 27 432 m (90 000 ft). Other design criteria regarding this vehicle are presented in Table 1. Possible missions for this type of aircraft include communications relay, reconnaissance and surveillance, standoff missile guidance, airborne early warning, atmospheric data collection, and drug interdiction.

The Lockheed Baseline HALE Aircraft features a primary joint at wing station 300.0 to per-

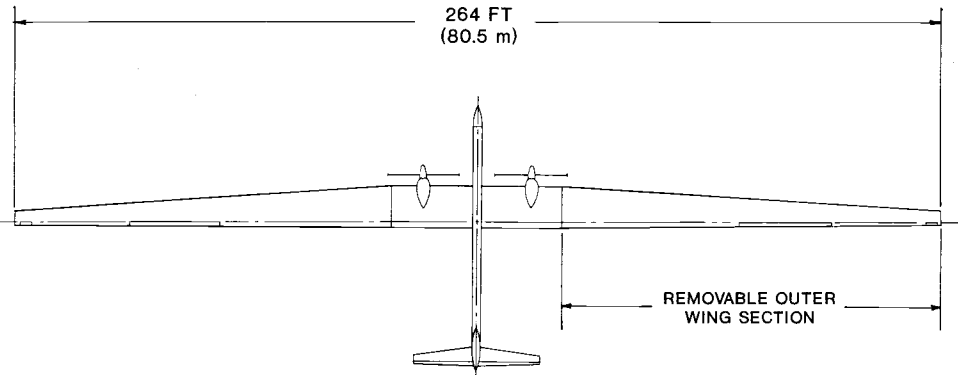


FIG. 5—Lockheed Baseline HALE Aircraft.

mit removal of outer wing sections. This wing station is just outboard of the engine nacelle and main landing gear. A traditional design approach would call for the use of titanium tension fittings at this location. However, the preferred joint design would be an all-composite, co-cured structure. This approach would eliminate the differential thermal expansion experienced in a typical metal-to-composite adhesive joint at the extremely low temperatures of high-altitude flight. Therefore, one criterion in the design of the stub-blade joint was a minimum use of metallic materials.

Critical loading of the Baseline HALE Aircraft wing was determined to occur during a gust condition at the end of the cruise flight segment. Accordingly, the stub-blade joint was sized to these critical loads for wing station 300.0. These design loads are presented in Table 2.

TABLE 1—Selected design criteria for representative Lockheed Baseline HALE Aircraft.

Maximum operating altitude, 27 432 m (90 000 ft)
Endurance, 1 to 3 days
Maximum airspeed, 111.1 km/h (60 KEAS) ^a
Design gross weight, 8163 kg (18 000 lb)
Wing span, 80.5 m (264 ft)
Maximum wing chord, 3.66 m (144 in.)
Wing aspect ratio, 29.0
Wing thickness ratio, 0.17
Wing taper ratio, 0.30
Gust load factor, N_z , 2.25
Flight maneuver load factors, N_z
Symmetric pullup, 1.30
Rolling pullout, 1.20
Pushover, -1.0
Taxi load factor, N_z , 1.25
Ultimate factor of safety, 1.25
Maximum wing tip deflection, 15% of semi-span
Thermal environment, -54.4 to +82.2°C (-130 to +180°F)

^aKEAS = knots equivalent airspeed.

TABLE 2—Baseline ultimate design loads for stub-blade wing joint at wing station 300.0 (gust cruise condition).

Spanwise bending moment, M_x , 790 854 N·m (7 000 000 in. · lb)
Axial load intensity (wing-box covers), $\pm 1\,389\,910$ N/m (± 7937 lb/in.)
Torsional bending moment, M_y , 16 947 N·m (150 000 in. · lb)
Vertical shear load, P_z , 55 600 N (12 500 lbs)
Note: M_z , P_x , and P_y are assumed to be negligible for the gust cruise condition.

Wing-Box Structure

The stub-blade wing joint design was based on the dimensional constraints presented by the Lockheed Baseline HALE Aircraft wing box. This wing box consisted of front and rear spars to carry vertical shear forces and upper and lower covers to react spanwise bending moments. The wing box also acts as a torque box to carry torsional loads. The front spar is located at 20.0% chord, and the rear spar is located at 45.0% chord, resulting in a wing-box width of 0.91 m (36.0 in.) for a 3.66-m (144.0-in.) total chord. The thickness-to-chord ratio is 0.17, resulting in a wing-box thickness of 0.62 m (24.5 in.). Ribs are located at 0.51-m (20.0-in.) intervals as dictated by spar cap stability requirements. Spars and wing-box covers use sandwich construction. Spars at wing station 300.0 consist of graphite/epoxy face sheets bonded to a 0.95-cm (0.375-in.) titanium honeycomb core. Wing-box covers at wing station 300.0 consist of graphite/epoxy face sheets bonded to a 1.27-cm (0.50-in.) titanium honeycomb core. Baseline HALE Aircraft wing structure and wing-box geometry are shown in Figs. 6 and 7.

Material Selection

Representative high-strength graphite/epoxy, high-modulus graphite/epoxy, and intermediate-modulus graphite/toughened-epoxy material systems were considered for use in stub-

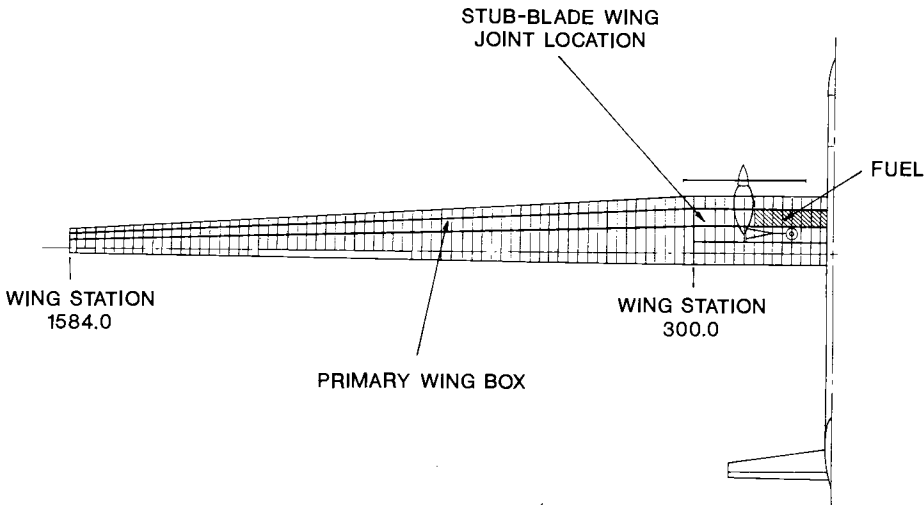


FIG. 6—Lockheed Baseline HALE Aircraft wing.

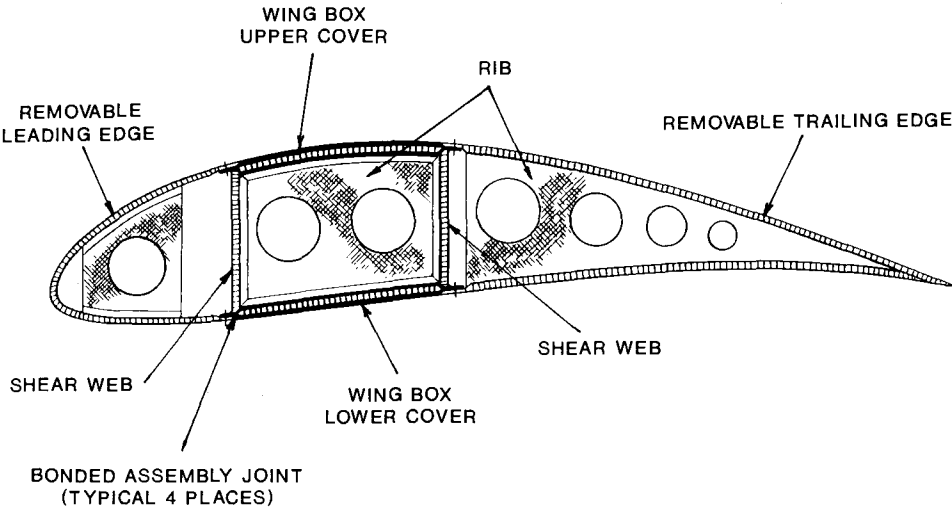


FIG. 7—Baseline HALE Aircraft wing-box structure.

blade joint design. Preliminary tensile and compressive laminate design allowables for these three material systems are shown in Table 3. A HALE aircraft must be able to withstand the extremely low temperatures encountered during high-altitude flight. Also, as stated previously, critical wing loading for the Lockheed Baseline HALE Aircraft was determined to occur at these extremely low temperatures during the critical descent segment. Therefore, estimated mechanical properties at -54.4°C (-130°F) were used in the material selection process. These properties were obtained by extrapolating actual ply-level test data and using standard methods for estimating laminate allowables. For ease of fabrication, it is assumed that the material

TABLE 3—Preliminary tension and compression design allowables for candidate wing-box materials estimated laminate properties at $T = -54.4^{\circ}\text{C}$ dry/wet [-130°F] [70/20/10 layup].

	HS/Epoxy ^a	HM/Epoxy ^b	IM/T-Epoxy ^c
$F_{x'}^{tu}$, MPa (ksi)	424.1 (61.5)	242.8 (35.2)	738.6 (107.1)
$F_{x'}^{cu}$, MPa (ksi)	527.6 (76.5)	375.2 (54.4)	680.0 (98.6)
$F_{x'}^t$, GPa (msi)	103.4 (15)	220.7 (32)	117.2 (17)
$E_{x'}^c$, GPa (msi)	103.4 (15)	220.7 (32)	117.2 (17)
$\epsilon_{x'}^{tu}$, μ strain	4100	1100	6300
$\epsilon_{x'}^{cu}$, μ strain	5100	1700	5800
ρ , g/cm ³ (lb/in. ³)	1.605 (0.058)	1.688 (0.061)	1.577 (0.057)
$E_{x'}\epsilon'/\rho^d$	1.060	0.577	1.585
$E_{x'}\epsilon'/\rho^e$	0.458	0.577	0.528

^a High-strength graphite/epoxy; T300/5208 or similar.
^b High-modulus graphite/epoxy; GY-70/930 or similar.
^c Intermediate-modulus graphite/toughened epoxy; T650-42/R6376 or similar.
^d Assumes allowable 1.0-g deflection ratio of 0.15.
^e Assumes allowable 1.0-g deflection ratio of 0.05.

used in the wing box proper will be the same material used in the stub-blade joint. Therefore, choice of a material for the stub-blade joint requires an understanding of the design drivers that dictate the choice of material for the wing.

Wing design philosophy is driven by stiffness and strength requirements. One requirement may take precedence over the other depending on aircraft performance and wing geometry. The objective in all cases, however, is to design the lightest possible wing and still meet the strength and stiffness criteria. Material selection is one key parameter for obtaining this minimum weight. Figure 8 illustrates the wing box ultimate fiber strain for various aspect ratios with the planform and wing thickness ratio of the Lockheed Baseline HALE Aircraft and a 1.0-g deflection ratio of 15.0%. Deflection ratio is defined as the wing tip deflection divided by the semi-span (that is, one half of the total span) of the wing. Any material with an allowable working strain to the left of the curve indicates a "strength-designed" wing. Additionally, any material with an allowable working strain to the right of the curve indicates a "stiffness-designed" wing.

The choice of the optimum material to achieve a minimum-weight wing, however, is not simply a choice of a high-strength material for a "strength-designed" wing or a high-modulus material for a "stiffness-designed" wing. The optimum material for least weight is the material with the highest specific working strength $E_x \epsilon' / \rho$, where ϵ' is the smaller of the strain as limited by the bending deflection stiffness criterion or the allowable laminate working strain. This parameter $E_x \epsilon' / \rho$ for all three materials is also presented in Table 3 and shows that the intermediate-modulus graphite/toughened-epoxy system (for example, T650-42/R6376 or a similar material system) has the highest specific working strength of the three candidate material systems for this specific design case. Therefore, this material was used in the design of the stub-blade joint. If the wing deflection criterion were changed, for example, to a 1.0-g deflection ratio of 5.0%, the intermediate-modulus/toughened epoxy system would not be the material of choice. Also, if several material systems exhibit comparable specific strengths for a specific design case, the choice of optimum material may be decided by additional factors such as cost and availability.

Large-span, high-aspect-ratio wings are, in general, stiffness critical to preclude wing divergence, control reversal (that is, opposite reaction to control input as a result of excessive torsional flexure), and torsion-bending coupling. The very low operating speed and resulting Mach number of the Lockheed Baseline HALE Aircraft have the beneficial effect of minimizing these stiffness requirements. The required torsional stiffness was determined using an iter-

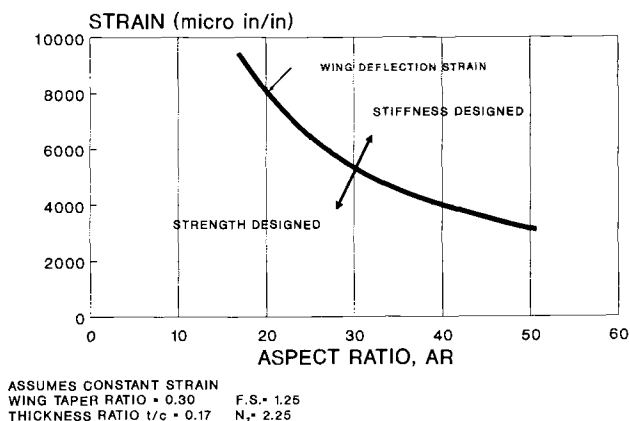


FIG. 8—Ultimate fiber strain versus aspect ratio for 1.0-g wing deflection ratio $D = 0.15$.

ative process of successive torsional stiffness distributions to determine the optimum distribution for minimum weight. Once this determination was made, the amount of biased ply material (that is, 45° ply sets) was established along the span of the wing box. Axial plies were then added to meet the spanwise stiffness requirement of a 15.0% deflection ratio at 1.0 g, resulting in an approximate 70/20/10 ply mix along the span of the wing. This 15.0% deflection ratio satisfied the flutter margin criterion and provided enough spanwise stiffness to meet dynamic taxi deflection limits. One of the interesting results of laminate stacking is that, when axial and biased ply sets are mixed in a wing cover, both stiffness quantities, AE^2 and Gt ,³ are greater than if axial plies were located exclusively in a “hard” spar cap and biased plies were located exclusively in a “soft” wing-box cover. Therefore, the Lockheed Baseline HALE Aircraft uses a wing cover that combines these axial (0°) and biased ($\pm 45^\circ$) ply orientations.

Design Approach and Philosophy

Every joint presents some degree of structural inefficiency with a resulting weight penalty. Therefore, joints should be avoided whenever possible in the design of lightweight aircraft structure. However, if a joint is required, every effort must be made to reduce this structural inefficiency and, thereby, minimize the weight penalty.

The stub-blade joint concept presents two primary sources of structural inefficiency: (1) an overlap of wing-box cover material, and (2) an overlap of shear beam (spar) material. These two elements are illustrated by the load distribution shown in Fig. 9.

The tensile and compressive cover loads in the overlap area decrease from a maximum value of $P_{\text{cover max}}$ to zero for both wing-box sections. Therefore, this joint incorporates excess weight if the amount of wing-box cover material in the overlap region is held constant. Fortunately, this situation is corrected easily in a composite material design by dropping off plies to match the loading diagram.

The second source of structural inefficiency, the overlap of shear web material, presents a more challenging problem. The apparent solution is to minimize the weight penalty by making the overlap length as short as possible. However, decreasing the overlap length increases the couple forces R_v required to resist the bending moment M_x . Therefore, there is a lower limit to a reasonable overlap length. This lower limit is defined by the prohibitive weight penalty imposed by the shear pin diameter and shear web thickness required to resist these increased couple loads. Couple loads can be decreased by lengthening the overlap, but an excessive overlap length also imposes a weight penalty on the joint. Therefore, design of a minimum weight stub-blade joint requires a scheme to optimize the blade overlap length.

Design of an efficient stub-blade joint also requires determination of which design approach is optimum for the blade. Blade designs can be grouped into two categories (Fig. 10): (1) blade designs that use pure shear beam action, and (2) blade designs that use pure truss action.

Both the pure shear beam approach and the pure truss approach present unique design problems. The pure shear beam design has a drawback in that the covers must include a substantial joggle for outer surfaces to be flush. This design also offers the maximum amount of shear web material for a given overlap length and requires close pre-alignment of the two wing-box sections for mating. On the other hand, the pure truss design eliminates the need for a shear web. However, this approach maximizes the kick loads (that is, balancing forces generated by eccentric load paths) at the blade-to-cover transition. There are also difficulties in designing a composite blade tip fitting so that pure truss action is achieved. Therefore, the stub-blade joint

² A = cross-sectional area of wing covers and E = elastic modulus of covers

³ G = shear modulus of covers and t = wing cover thickness
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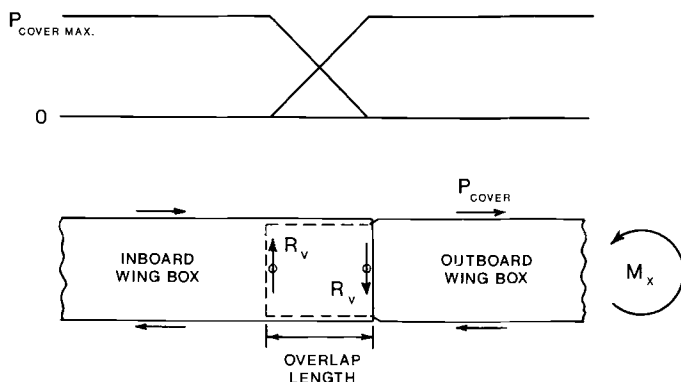


FIG. 9—Cover load distribution for stub-blade type joint.

presented in this paper uses a combination of the two design approaches. The blade is tapered to take advantage of partial truss action and also provide ease of mating. Shear webs are included to provide a location for shear pin bushings and bearing material and provide a direct path for shear loads not reacted by truss action. Also, the angle of the blade in relation to the wing box can be specified so that kick loads at the blade-to-cover transition are minimized.

Blade Length and Blade Angle Optimization

Optimum blade length was selected based on results of the Stub-Blade Joint Optimization Program (SBJOPT). This FORTRAN program calculated the weight of each component of the stub-blade joint based on the pin-to-pin dimension "D2" and the resulting couple forces required to resist the spanwise bending moment M_x . The effect of the vertical shear load P_z and torsional moment M_t was also considered in this program, but the magnitudes of the internal joint forces resisting these loads are independent of the dimension "D2." Therefore, these loads only affect the total weight of the joint and not the optimum blade length. The program

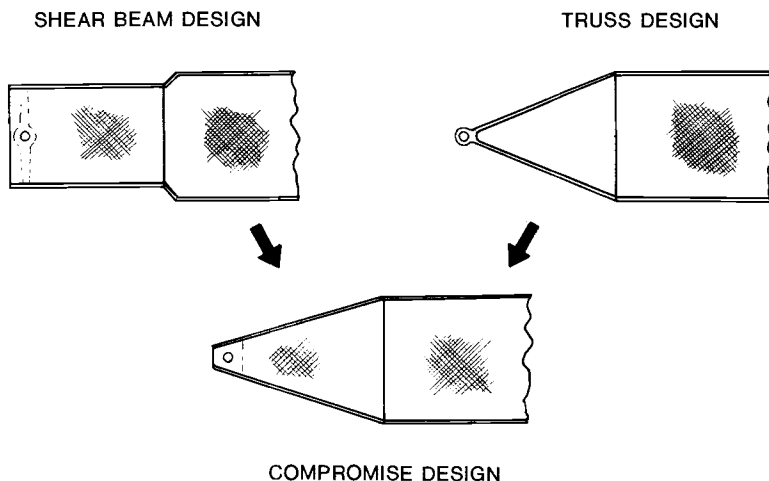


FIG. 10—Design approach for blade structure.

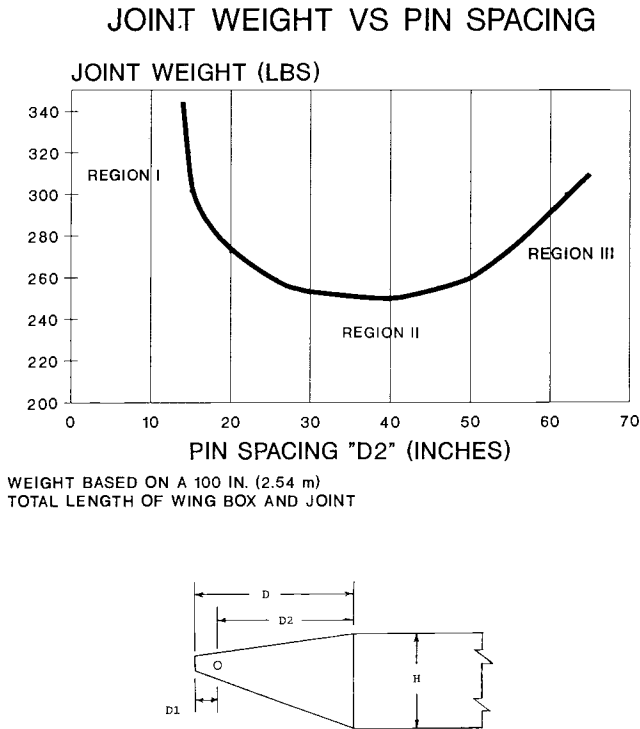


FIG. 11—Plot of *SBJOPT* program output (total joint weight versus pin spacing "D2").

sums all component weights to arrive at a total joint weight. This total joint weight is based on a 2.54-m (100-in.) section of wing box. A plot of total joint and associated wing structure weight versus pin spacing dimension "D2" is shown in Fig. 11. Region I of this plot illustrates the weight penalty imposed by prohibitively large shear pins, pin support structure, and shear webs. Region III illustrates the weight penalty caused by excessive shear material overlap. Region II represents a zone of optimum overlap and indicates that the total weight of the joint is minimized when $D2 \approx 1.02 \text{ m} (\approx 40 \text{ in.})$. Therefore, this dimension was incorporated into the preliminary stub-blade joint design.

Kick loads at the transition of the wing-box cover to blade cover were minimized by adjusting the angle of the blade in relation to the wing box. These kick loads are minimized when the assembly pin is located approximately 19.68 cm (7.75 in.) below the upper surface of the wing box. This optimum blade angle is applicable only for the specific case of the 2.25-g maximum wing upbending (gust condition) and 1.0-g maximum wing downbending (pushover/negative gust) load factors generated for the Lockheed Baseline HALE Aircraft.

Detail Design and Structural Analysis

Lockheed-developed composite analysis programs as well as other closed-form solutions were used to size shear webs and tensile and compressive covers for strength and stability. Design of all components and fittings was driven by the requirement for lightweight structure using a minimum of metallic materials. Certain loadings and load intensities, however, dictate the use of metals for certain components. Use of metallic materials was limited to fixed shear pins and assembly pin (AM-355 stainless steel), self-aligning spherical bearings (440C stainless

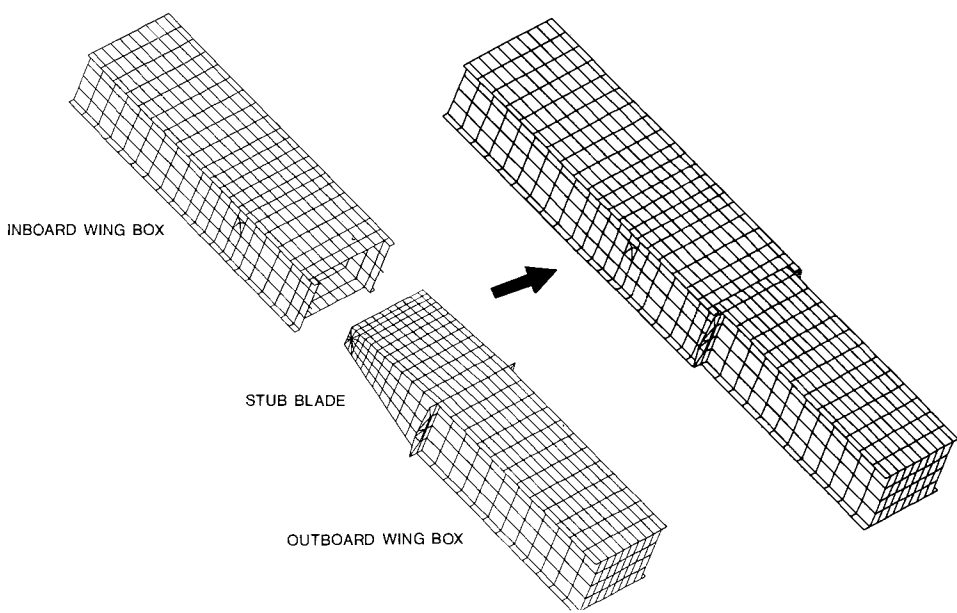


FIG. 12—NASTRAN finite-element model of stub-blade wing joint.

steel inner ring and 410 stainless steel outer ring), and bushings (Ti-6Al-4V alloy outer ring and beryllium copper inner ring). A NASA Structural Analysis (NASTRAN) model (Fig. 12) was then constructed based on this preliminary design and used to gain more insight into the internal load distribution of the joint. Based on the results of this model, the initial design was revised, resulting in the final stub-blade wing joint shown in Figs. 13 and 14.

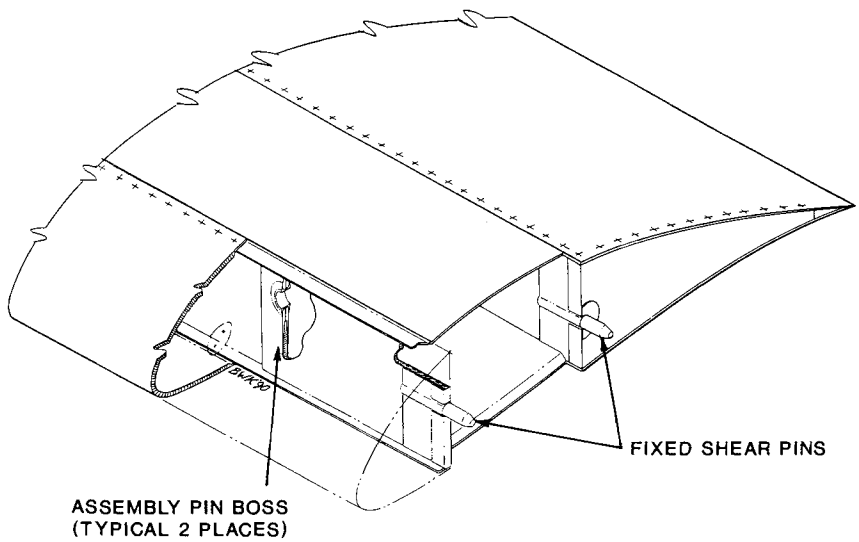


FIG. 13—Stub-blade wing joint design (inboard section).

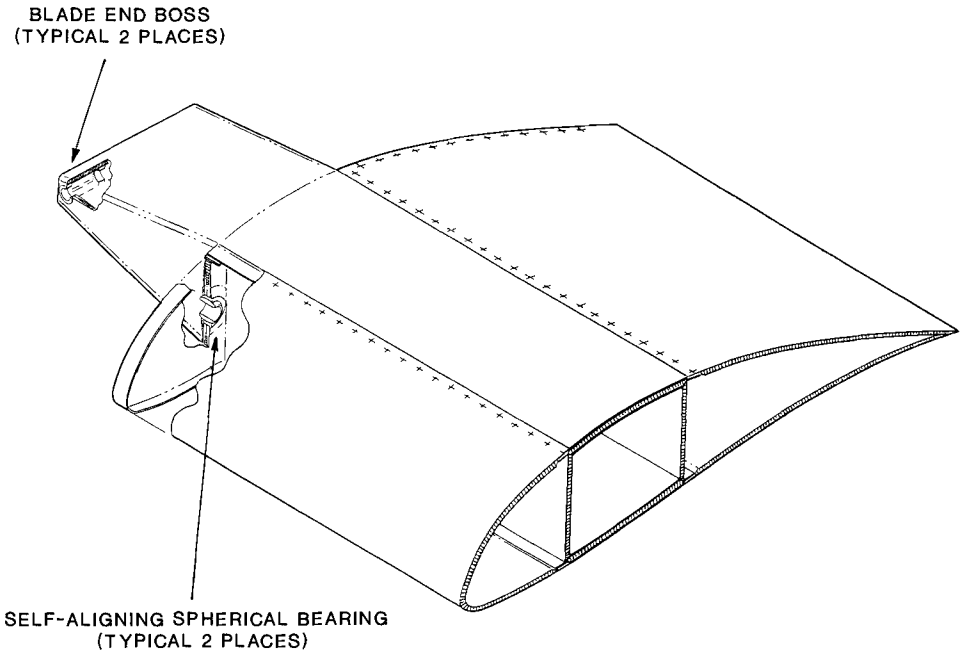


FIG. 14—Stub-blade wing joint design (outboard section).

Four primary fittings within the stub-blade joint are of interest. These are: (1) the assembly pin boss (inboard wing box), (2) fixed shear pins, (3) blade end boss, and (4) self-aligning spherical bearings. The main assembly pin boss located in the shear web of the inboard spar transfers transverse shear (P_z) and axial load (P_y due to centrifugal force). These loads bear on a beryllium copper bushing press-fit into a split titanium housing, which bears on a full-depth laminate. Each split housing has a large tapered external flange, which is bonded and trapped by the sandwich spar web face sheet. A reinforced region over the full height of the spar provides column stability for the shear transfer and beam strength for axial load transfer to the wing-box covers.

The fixed shear pins on the inboard side of the wing joint react couple loads resulting from wing bending and fore-aft drag (P_x) loads. The drag load is reacted by the closeout rib through bearing. The more significant internal couple load (R_v) is applied directly to the spar webs. The metallic pin is trapped in a woven preform with integral webs that are bonded directly to the shear web face sheets. Internal to the sandwich are composite tubular columns, which are loaded through bearing-push action from the shear pin. Strain compatibility must be ensured on the “pull” side. The assembly flange of the spar-to-cover joint is increased locally in area to support initial box bending until the covers become effective. The spar web sandwich between the assembly pins was increased in thickness to be more compatible with the large-diameter fixed pins and increased web shear load. The main assembly pin receptacle on the tip of the outboard blade supports the assembly pin in a similar fashion to that of the inboard spar web. Two design considerations are of importance. The first is that additional pin support must be provided internally at the blade tip. This internal support forces the blade covers to stiffen the assembly pin and reduces pin bending caused by the gap created between the inboard shear web and the blade edge. This gap was created to provide clearance for the fixed shear pins on the inboard wing-box side during wing assembly. The resulting pin rigidity fur-

ther reduces the secondary bending (M_y) applied to the spar shear web via the fixed shear pins. The second design consideration is that the highly loaded shear path between the blade assembly pin bushing and closeout rib spherical bearing must be kept a jointless, uninterrupted load path. The kick loads generated by turning the blade web shear loads in a chordwise direction are reacted by closeout rib bending. The self-aligning spherical bearings have an integral machined mounting flange, which transfers load through bearing and flange shear.

It is envisioned that mating of wing sections using the stub-blade joint would involve supporting the outer wing section on two adjustable carts. These carts would allow fine adjustment of the elevation and angle of the outer wing as it is brought to mate with the inner wing section.

Comparative Tensile Joint

A traditional tension-type joint was designed to the same dimensional constraints imposed by the preexisting Lockheed Baseline HALE Aircraft wing design and sized to the same design loads as the stub-blade joint. This tension joint allowed a basis for comparing the relative merits and drawbacks of both wing joint designs. The comparative tension joint design is shown in Fig. 15. Tension fittings are constructed of Ti-6Al-4V alloy. Flush mechanical fasteners were used to join these tension fittings to the graphite/epoxy wing-box covers. Mechanical fasteners were required because the extreme temperature ranges experienced by a HALE vehicle would result in considerable differential thermal expansion between the metallic and composite materials, thereby precluding the use of a bonded joint at this critical location. Also, the tension joint was designed to the constraint that the outer mold line (OML) of the wing-box surface could not be exceeded. In other words, the tension joint design was required to be flush with the aerodynamic surfaces of the wing box, preventing location of the tension bolts on the wing-box cover neutral axis. The justification for this requirement is that HALE aircraft performance at 27 432 m (90 000 ft) demands that disturbance of the airflow over the wing be held to a minimum.

Shear blocks were used to transfer vertical shear loads from the tension fittings to the spars via the close-out rib. These shear blocks also react the local bending moment generated by the eccentricity of the tension bolts in relation to the wing-box cover neutral axis. A direct shear splice was not used to transfer shear loads from the outboard to inboard spars because this type

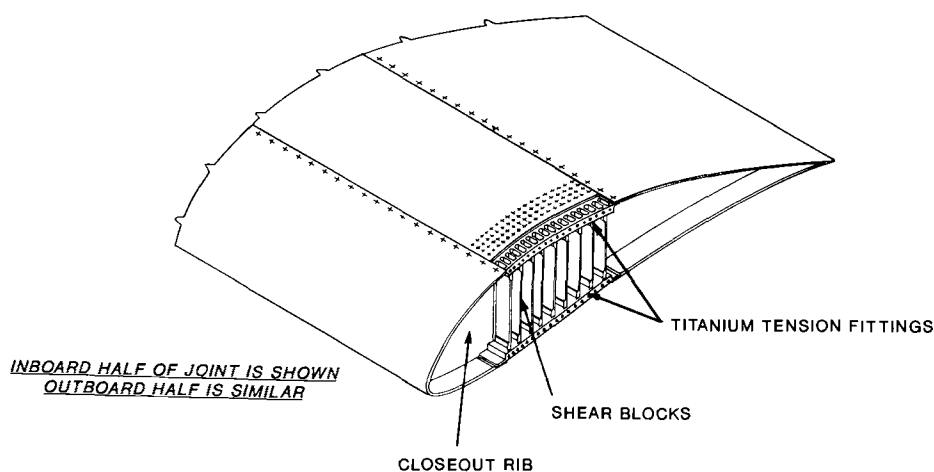


FIG. 15—Comparative tensile-type wing joint.

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of connection requires very close alignment of fastener holes and is not conducive to repeated assembly and disassembly

Note that removal of an outer wing section using this particular joint design requires the removal of 36 tension bolts, 18 for the upper cover and 18 for the lower cover. The number of tension bolts was based on the best compromise between minimizing eccentric loading on the tension fittings caused by larger bolts and a practical minimum tension bolt size and number required to transfer wing loads. This relatively large number of tension bolts introduces potential alignment problems and, in general, makes assembly of wing sections difficult and time consuming.

Weight Comparison

A total weight was calculated for the stub-blade joint and the comparative tension joint. Both weight build-ups consider a 1.27-m (50.0-in) long section of wing box. This preliminary weight comparison indicates that the stub-blade concept is weight-competitive with a traditional tension joint. The estimated weight for the stub-blade joint and associated wing-box structure is 74.8 kg (165 lbs). The estimated weight for the comparative tension joint and associated wing-box structure is 68.5 kg (151 lbs), or 8.5% less than the weight of the stub-blade joint.

Conclusions and Future Work

The work presented in this paper has shown that, compared to a traditional tension type wing joint, the stub-blade wing joint is

- A viable alternative design that warrants further development and consideration for HALE aircraft,
- Weight-competitive,
- Easily assembled and disassembled with a single pin,
- A practical design that uses the advantages offered by composite materials (that is, bonded and co-cured structure, tailoring of mechanical properties),
- A design with few metallic components, thus minimizing differential thermal expansion problems,
- An aerodynamically smooth joint devoid of “bumps” or protruding fairings, and
- A concept free of “black aluminum” mentality

The true test of any composite design, however, lies in actual fabrication and structural testing of hardware. Future development of this concept will initially require subcomponent testing of the fixed shear pin design in addition to testing of the assembly pin boss and self-aligning spherical bearing support structure. These subcomponent tests must be performed at the extremely low temperatures of high-altitude flight so that the differential thermal expansion and brittle behavior of all materials can be accurately evaluated.

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Temperature and Load Cycling of a Thick Polyimide Quartz Laminate

REFERENCE: Laurencot, R. E. and Keen, L. M., "Temperature and Load Cycling of a Thick Polyimide Quartz Laminate," *Composite Materials Testing and Design (Tenth Volume)*, ASTM STP 1120, Glenn C. Grimes, Ed., American Society for Testing and Materials, Philadelphia, 1992, pp. 279–287

ABSTRACT: This paper presents the test results of temperature and load cycling of a thick polyimide quartz laminate. The specimens tested were all of the same geometry, a V-shaped coupon with an apex radius of 1.0 in (2.54 cm), width of 1.0 in (2.54 cm), and a thickness of 0.42 in (1.07 cm). The load cycle is applied by creating a moment at the apex of the V of 52.5 in-lb (5.95 Nm). For each load cycle, there are three thermal cycles, each consisting of a 10-min ramp up from room temperature to 550°F (288°C), a constant 20-min period at 550°F (288°C), and a 10-min ramp down from 550°F (288°C) to room temperature.

The V-shaped specimens were tested for ultimate bending strength at various times through the testing. Some specimens were tested before cycling began, and others were tested at convenient intervals during the cycling. Measurements were taken of the weight changes, thickness changes, and ultimate bending moment decreases. Plots of the relationships of physical properties to load/thermal cycling are presented as well as photomicrographs indicating effects of thermal cycling on the resin/fiber system.

From the test results, an empirical procedure for estimating the life of a thick laminate, subject to a similar environment, can be extrapolated.

KEYWORDS: thick laminates, polyimide quartz, temperature cycling, load cycling, service life, loss of properties

This paper presents the test results of temperature and load cycling of a thick polyimide quartz laminate. The purpose of this test program was to determine the effects of such cycling on the physical and mechanical properties of the laminate. Once this was determined, these effects could be estimated and a service life could be predicted for a structure in a similar environment.

Materials and Processes

The laminate was fabricated from Astroquartz 581 fabric plies and BPI-373 condensation polyimide matrix. All specimens were made by Brunswick Defense. The laminate consisted of 44 plies, resulting in an average laminate thickness of 0.42 in (1.07 cm). Each layup was B staged and then held at 360°F (182°C) in an autoclave cure, at positive pressure, for 2 h.

After the final layup was applied, the specimens were postcured at 600°F (316°C) in the autoclave. All specimens were ultrasonically inspected after postcure to ensure there were no unacceptable defects.

¹Group engineer and stress engineer, respectively, Advanced Development Projects, Lockheed Aeronautical Systems Company, D/72-37, B/375, P/B-6, Burbank, CA 91520

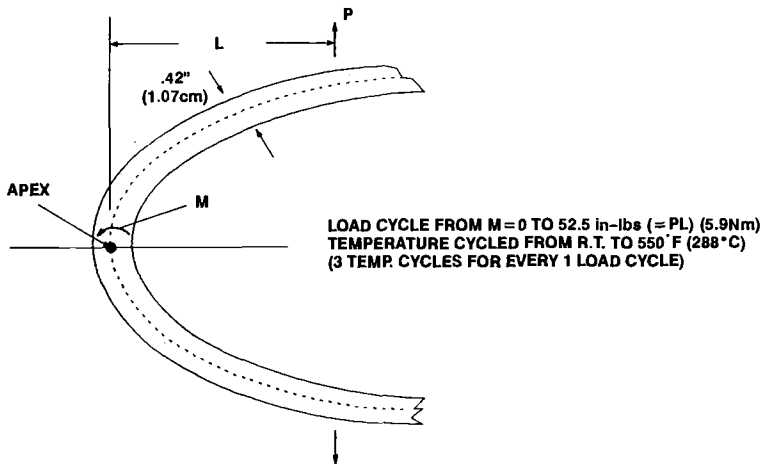


FIG. 1—Test specimen.

Description of Testing

All the specimens tested were of similar geometry, a V shape loaded in such a manner as to produce maximum bending moment at the apex (Fig. 1). There were 18 specimens in the test. The internal stresses at the apex of the V are shown in Fig. 2. All specimens were 1.0 in. (2.54 cm) in width and formed to a 1.0-in. (2.54-cm) radius at the outside mold line (OML). Thicknesses of the specimens averaged 0.42 in. (1.07 cm). The length of the specimens varied slightly (see “L” in Fig. 1) but averaged close to 5.0 in. (12.7 cm). The test load was adjusted so that the ending moment at the apex was always 52.5 in. ·lb (5.9 Nm). The shape of the specimen was selected so that the maximum bending moment and flatwise tension stresses occurred at the same point. The arrangement of test equipment is shown in Fig. 3. The recording and

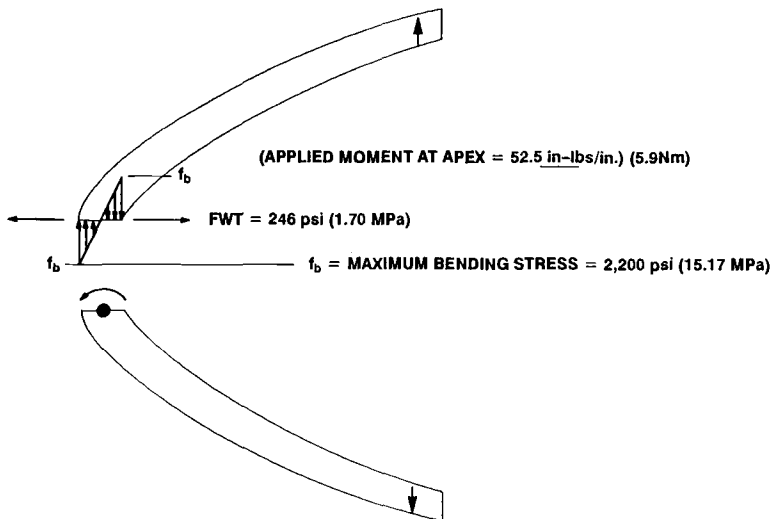


FIG. 2—Stresses in laminate.

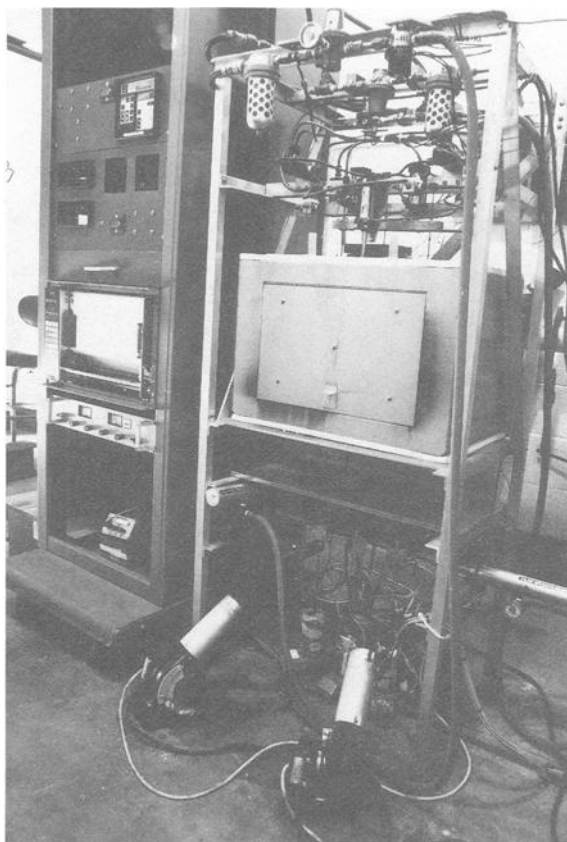


FIG. 3—Test equipment arrangement.

monitoring equipment is on the left side of the photograph. Figure 4 shows three specimens mounted in the oven.

The temperature cycle consisted of a ramp up from room temperature (RT) to 550°F (288°C) over 10 min; a constant dwell at 550°F (288°C) for 20 min; and a ramp down to RT over 10 min. This thermal cycle was repeated three times while the load cycle was maintained at a constant 52.5 in.·lb (5.9 Nm). This load is applied at the beginning of the first thermal cycle and remains for three thermal cycles. At the end of the third thermal cycle, the load is reduced to zero. Figure 5 shows a diagram for a typical temperature cycle. To evaluate the relative effects of load versus temperature cycling, one specimen was only thermally cycled for 1833 h without any load while the other two adjacent specimens were both thermally and load cycled (see Figs. 1 and 5). The unloaded specimen exhibited the same general physical effects that the loaded specimens exhibited and fell within the scatter band of the loaded specimens. Consequently, it was assumed that the structural load on the specimens was of minor significance in relation to the thermal load for the particular stress level used in these tests. The uncycled ultimate bending strengths of the specimens used in these tests varied from approximately 400 to 700 in.·lb (45 to 79 Nm) at 550° F (288°C). The differences in strengths are attributed to the fact that the specimens came from different layups fabricated over a period of several

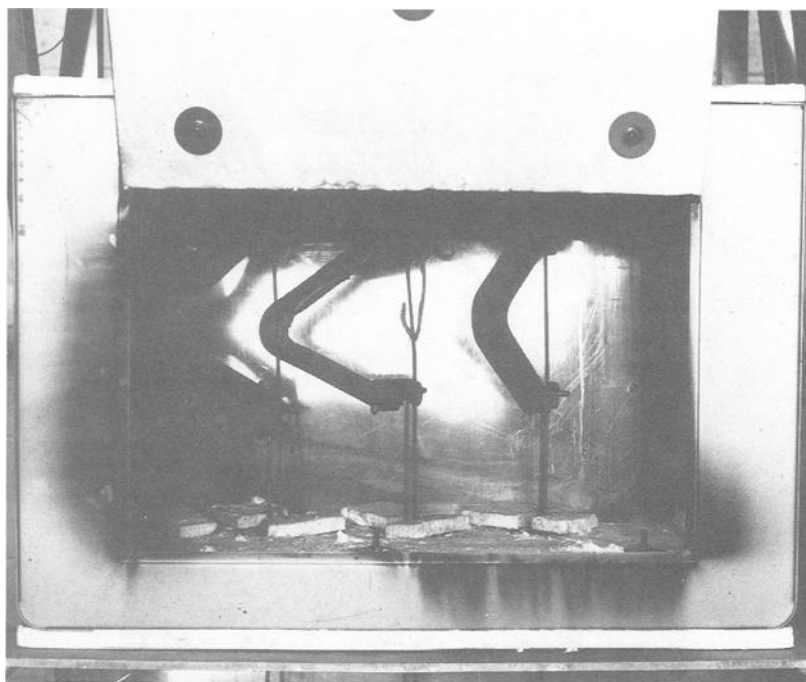


FIG. 4—Specimens mounted in oven.

Test Results

Figure 6 shows the laminate cross section before and after load and thermal cycling. The porosity increased from 10 to 33% after 1833 h at 550°F (288°C). The porosity was measured by image analysis (area) method. It is assumed that this represents typical volumetric porosity even though it is a cross-sectional measurement. Exposure to 550°F (288°C) is considered very severe for a polyimide quartz laminate. Figure 7 shows a typical flatwise tensile (FWT) type failure in the radius section before cycling (note the thin white line across the center of the photograph) and the typical laminate section from the same area after cycling. Figure 8 is a more detailed view of the radius section of the laminate before and after cycling. Figure 9 shows the well compacted and wetted fibers before cycling and the matrix microcracking after 1833 h at 550°F (288°C). Figure 10 shows a scanning electron microscope (SEM) photomicrograph of a failure surface before and after cycling. The matrix hackles of a typical laminate failure can be seen adhering to the fibers on the uncycled specimens. The cycled specimen shows relatively smooth fibers with little matrix adherence. Figure 11 shows enlarged SEM views of the surfaces before and after cycling. The SEM taken before cycling shows the details of the fiber-matrix bonding along a length of a typical fiber; the SEM taken after cycling shows the loss of matrix and possibly the sizing on the fibers.

The quantitative results of this testing are presented in Figs. 12 through 15. The relationship of specimen thickness loss to time at 550°F (288°C) is shown in Fig. 12 and is 0.0108 in. (0.0274 cm) decrease in thickness per 1000 h at 550°F (288°C). This represents a thickness loss of 2.44% per 1000 h. The thicknesses of each specimen were measured at a minimum of five locations.

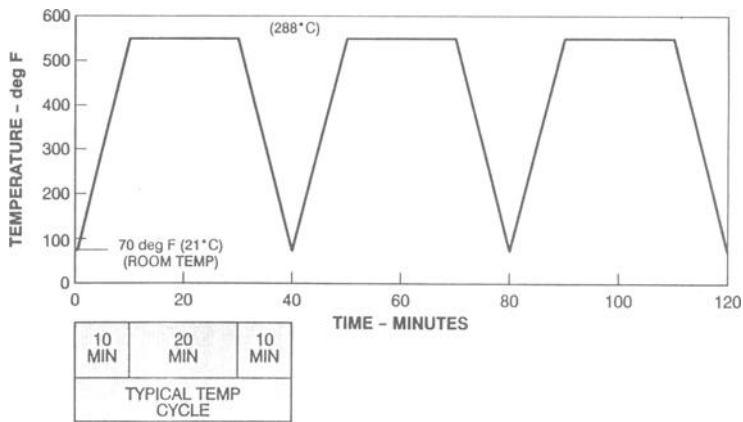


FIG. 5—Typical temperature cycle.

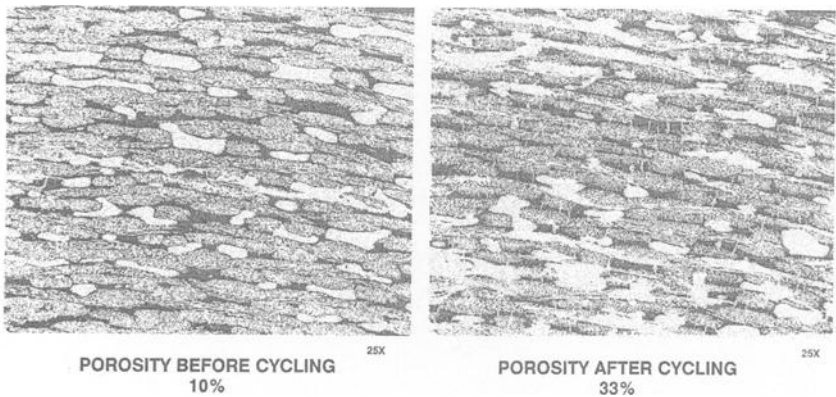


FIG. 6—Laminate porosity.

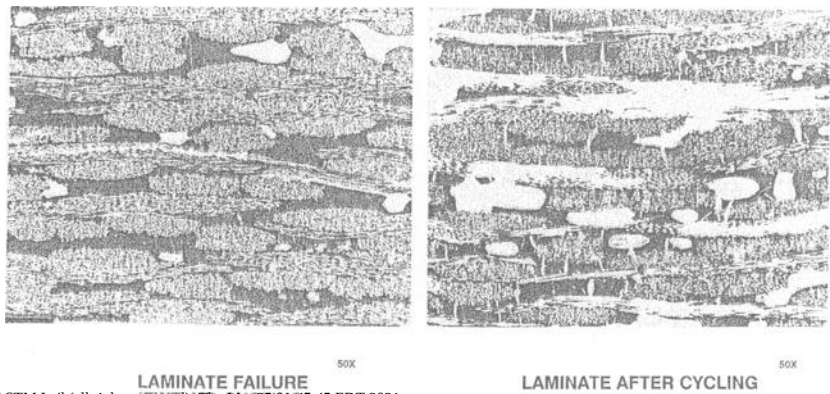


FIG. 7—Laminate failure and microcracking at $\times 50$.

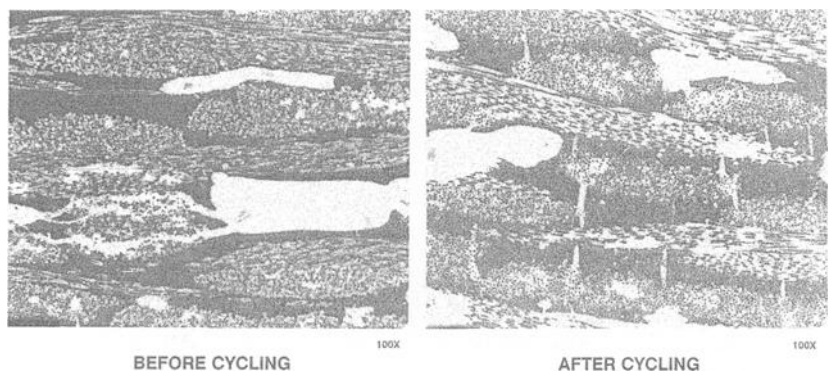


FIG. 8—*Laminate microcracking at $\times 100$.*

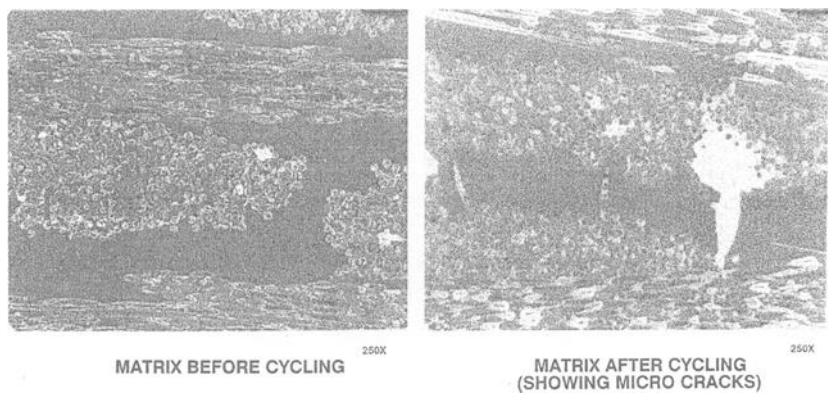


FIG. 9—*Laminate microcracking at $\times 250$.*

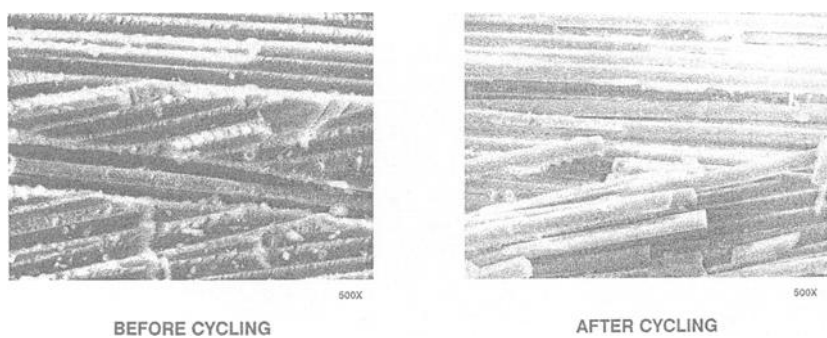


FIG. 10—*Laminate matrix degradation at $\times 500$.*

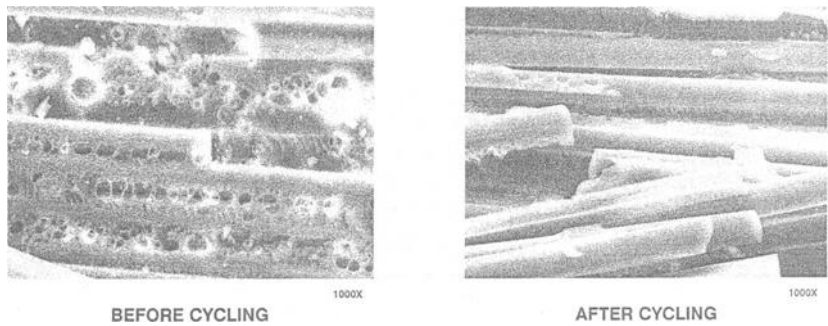


FIG. 11—Laminate matrix degradation at $\times 1000$.

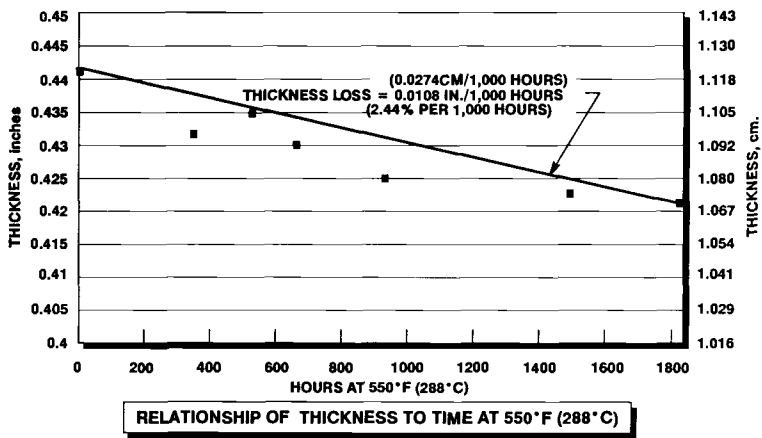


FIG. 12—Relationship of thickness to the time at temperature.

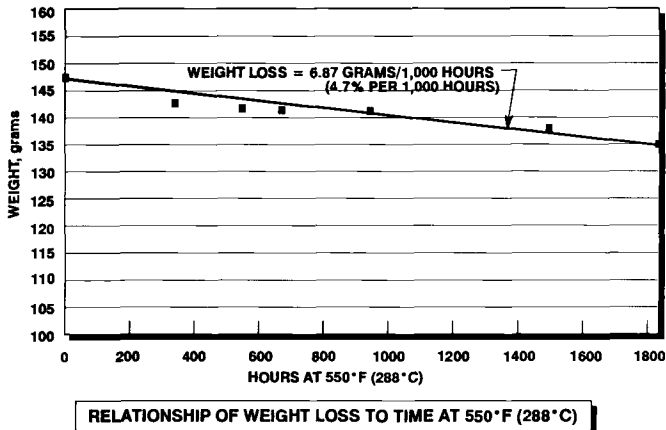


FIG. 13—Relationship of weight loss to the time at temperature.

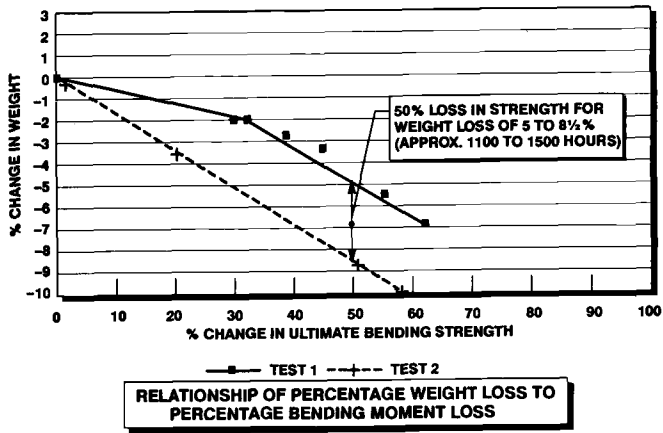


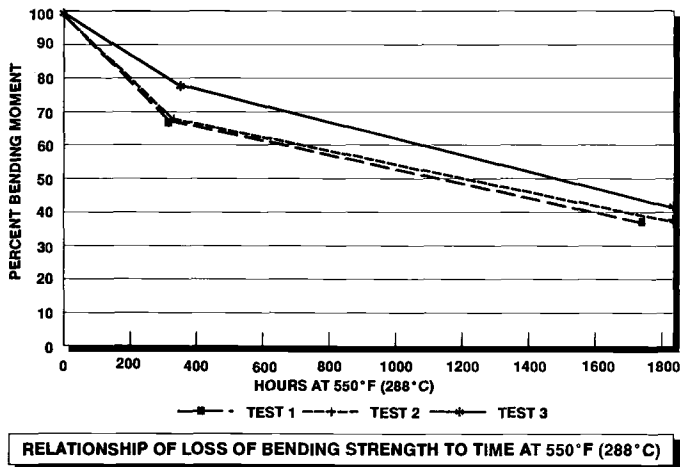
FIG. 14.—Relationship of weight loss to bending moment loss.

The relationship of specimen weight loss to time at 550°F (288°C) is shown in Fig. 13 and is 6.87 g per 1000 h. This represents a weight loss of 4.7% per 1000 h. The weight loss is attributed to the loss of matrix and possibly sizing in the interphase region. Figures 12 and 13 represent the averages of specimens from three different manufacturing layouts.

An envelope of the percentage weight loss to the percentage bending strength decrease is shown in Fig. 14. All other test data fell within the two curves shown and resulted in a weight loss range of 5 to 8.5% for a 50% decrease in strength for approximately 1100 to 1500 h at 550°F (288°C). These two curves represent individual test specimens.

The test conditions for the data presented in Figs. 12 through 15 are as described in Fig. 1 for structural load combined with Fig. 5 for thermal load.

Figure 15 depicts the relationship of loss in bending strength to hours at 550°F (288°C). All test specimens fell within the limits shown on the graph. With this relationship established, it is now possible to estimate the service life of a structure thermally cycled as shown in Fig. 5.



For example, if a given structure requires an ultimate bending strength of 300 in. lb (33.9 Nm) and its uncycled strength is 500 in. lb (56.5 Nm), the service life of the structure is approximately 700 h, if arbitrarily based on the lowest curve shown in Fig. 15 (that is, required bending moment is $300/500 = 60\%$). The three curves shown in Fig. 15 represent individual specimens from three different manufacturing layups.

Because of the method of testing, moisture effects on the laminate were eliminated from the above results. For structures that experience conditions in which moisture is absorbed by the laminate and is subsequently dried by undergoing the described thermal cycling used in these tests, an appropriate knockdown factor should be used in determining service life.

Conclusions

The following conclusions may be drawn as a result of this test program of thermal and load cycling:

1. The matrix is subject to oxidation as reflected in Fig. 13.
2. Degradation of the interphase occurs as shown in the photomicrograph and the scanning electron microscope photographs (Figs. 9 through 11).
3. Mechanical properties are reduced. The thickness of the specimens decreased at a rate of 2.4% per 1000 h at temperature. The weight of the specimens decreased at a rate of 4.67% per 1000 h. The bending strength of the specimens decreased by 50% in approximately 1100 to 1500 h at 550°F (288°C) as shown in Fig. 15. These reductions are attributed to the loss of matrix due to oxidation and the degradation of the interphase region.
4. The thermal load on the specimens was the primary cause of the physical and mechanical property losses. The internal stresses imposed by the bending moment test load had a negligible effect.
5. The service life of a polyimide quartz laminate, subject to the same load and thermal cycling used in this test program (Figs. 1 and 5), can be predicted by use of Fig. 15, as long as the uncycled bending strength has been established.

Session VI

Quality Assurance and Process Control

Karl V. Steiner, Chairman

Overview

Quality assurance and process control often go hand-in-hand. The information gathered by the quality control specialist is of great importance to the people monitoring and designing process control parameters. On the other hand, each change of control parameters will influence the performance of the structure and should therefore be investigated by the quality assurance division. In the composites industry, this interaction is even more crucial since many processing issues are still new and not completely understood.

The study of the "Effects of Processing Variables on the Quality of Co-cured Sandwich Panels" by Jouin, Pollock, and Rudisill is one such example. A number of material systems are subject to a variety of processes. The resulting sandwich quality is determined through a number of destructive test procedures such as leak, compression, and tension testing. This study eventually leads to improved performance of the sandwich panels by optimizing the lay-up sequence, the cure cycle, and the choice of surfacing plies.

"An Evaluation of the Current Status of Automated Process Control for Thermosetting Composites" by Holl and Rehfield focuses on another side of the issue. In this case, in-process control is used to provide information about the physical state of the resin during the cure cycle. Transducers based on dielectric, acoustic, and fluorescent signals are used as sensors in a feedback loop. They can be interfaced with algorithms which adjust the temperature and pressure profiles during the cure cycle. In-process control is an important area of research since fluctuations during the cure cycle have a direct impact on the final performance of the composite structure. Capabilities, cost, and size of available transducers are crucial parameters that have to be taken into account when selecting the proper in-process monitoring system.

The "Analysis of Composite Materials Containing Defects" by Frankle, Jones, Roberts, and Shusto using nondestructive evaluation techniques highlights the advantages of quality control methods combined with advanced design tools. This study shows a dramatic step forward in the interaction of design and analysis tools used to predict the performance of actual composite parts. A finite element structural analysis code is used to model the performance of carbon-carbon rings. Digital ultrasonic NDE data are then superimposed to develop a mechanism for automatically including the effect of defects in the analysis of the actual composite structure. Changes in matrix- and fiber-dependent properties based upon the ultrasonic data are taken into account in the analytical techniques. Subsequent hoop-tension tests have shown good correlation between the finite-element code predictions and the actual performance of the rings.

"Image Enhancement Techniques for Ultrasonic NDE Applications" by Steiner have proven to be a reliable method for the inspection of composite structures. One of the barriers is the dissemination of information from the quality control department to the manufacturing and design department. Image enhancement techniques allow the analyst to improve dramatically the detectability of defects and visibility of critical information gathered about the internal structure of the composite. A variety of techniques such as contrast stretching, histogram equalization, Kernel enhancement, and fast Fourier transformations can be applied. These techniques, combined with statistical histogram analysis and digitized full-volume waveform analysis, are powerful tools for the classification of flaws. Classification of defects

into such categories as porosity, contamination, delaminations, and matrix-rich or matrix-starved regions and information about the flaw depth provide important information which can be used by both design and manufacturing to improve the quality of subsequent production runs.

This session covers the effects of process variables on co-cured sandwich panels, automated process control for thermosetting composites, the analysis of defects in composites using non-destructive evaluation techniques, and ultrasonic NDE techniques as a reliable way of inspecting composite structures

Effects of Processing Variables on the Quality of Co-cured Sandwich Panels

REFERENCE. Jouin, P , Pollock, D , and Rudisill, E , “Effects of Processing Variables on the Quality of Co-cured Sandwich Panels,” *Composite Materials Testing and Design (Tenth Volume)*, ASTM STP 1120, Glenn C. Grimes, Ed , American Society for Testing and Materials, Philadelphia, 1992, pp 293–307

ABSTRACT: This paper summarizes basic studies conducted to compare the effects of various processes and materials on the quality of co-cured honeycomb sandwich panels with carbon-epoxy facings. Manipulation of the vacuum-pressure-temperature schedules of those processes resulted in improved facing surface quality, compaction, and sealing. The mechanical properties of those panels also improved.

KEYWORDS: carbon fiber composites, epoxy matrix, honeycomb core construction, co-cured process, quality evaluation, mechanical testing

The cost advantages realized by co-curing honeycomb core panels with composite skins are offset by a reduction in overall quality [1]. Typical problems encountered are core movement, bag side skin dimpling, surface porosity, interlaminar voids, and panel leaks. For external aircraft structures, these problems can result in lower strength. Skin delamination could also occur when the skin voids fill with water while the aircraft is on the ground, which then freezes when the aircraft is in flight.

A test program was designed to measure the effect of process variables on the quality of the facings of honeycomb sandwich panels. Quality was measured by surface finish on the tool side, facing compaction, facing strength, and the ability of the panel to be leak-free.

Process Description

Production sandwich panels are molded on carbon-epoxy tools. A carbon fabric preimpregnated with an epoxy resin is used to laminate the panel facings. The core is a honeycomb Nomex® or glass paper. A film adhesive is applied between core and facing. The cure is performed in an autoclave, with the baseline process following this cure profile: Apply 28 in Hg (711 mm Hg) minimum vacuum, apply 20 to 30 psi (138 to 207 kPa) pressure, vent vacuum, raise temperature to $250 \pm 10^\circ\text{F}$ ($121 \pm 6^\circ\text{C}$) at 1 to 3°F/min (0.5 to 1.7°C/min), and cure for 2 h at $250 \pm 10^\circ\text{F}$ ($121 \pm 6^\circ\text{C}$).

This process should allow the following events to occur concurrently: compaction of the laminate, adhesive flow to form fillets to bond facings to the core, and controlled resin flow to fill interlaminar voids and pinholes left during the impregnation of the carbon cloth.

¹Materials engineers, McDonnell Douglas Helicopter Company, 5000 E. McDowell Rd., Mesa, AZ 85205

The process variables that can affect the quality of the facings studied in this work are:

- vacuum level, pressure level and schedule;
- temperature schedule including intermediate hold;
- debulking schedule during the lamination of the skin; and
- precuring the skin.

Other variables included:

- fabric weave,
- core density and cell size, and
- resin content.

Test Program

The scattered porosity found on the tool surface of production honeycomb sandwich panels indicated that an underlying problem existed that affected the laminate through the thickness. Subsequent leak testing of the panels and cross sections of the skin confirmed that the problem was not confined to the surface of the skins, but that the skins on the tool side and on the bag side were porous and were not properly compacted (Fig. 1). The cross section showed that the laminate plies, which had been compacted or debulked during lay-up, would separate between the honeycomb walls during processing, thus opening a network of interconnected voids between the outer surface and the core space.

This condition, associated with the cure of epoxy resins, has led others to modify their process [1,2]. It was therefore decided to correct this condition by modifying either the process or the material or both. A test program to select the best approach was initiated. This program included the following elements:

- (1) design standard test panels,
- (2) list all process or material modifications that could affect the quality of the panels,
- (3) fabricate the panels and screen the candidates by conducting leak testing and compaction studies,
- (4) carry out mechanical tests on the best panels, and

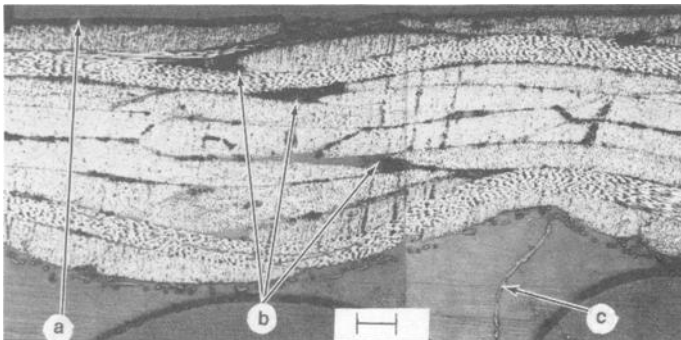


FIG. 1—Cross section of pore detected via leak testing: (a) tool surface, (b) voids, (c) honeycomb cell wall. Scale marker indicates 250 μm .

- (5) verify results by fabricating full-scale production panels using the most promising approach.

This study only reports the results of testing on small panels.

Standard Test Panel Configuration

Geometry

The test panels were designed to represent a typical production sandwich panel. Details incorporated included: close-out configuration and core bevel angle, flange width, fabric orientation, ply-laminating sequence, and supported film adhesive between facing and core. Core thickness was standardized at 1 in. (25.4 mm). Panel width was set at 18 in. (457 mm) (see Fig. 2).

Materials and Fabrication

Materials qualified for production were used. Fabrication followed production process specifications with the deviations noted below. A flat 0.25-in. (6.35-mm) thick carbon/epoxy caul plate, simulating a production tool, was used to mold all panels.

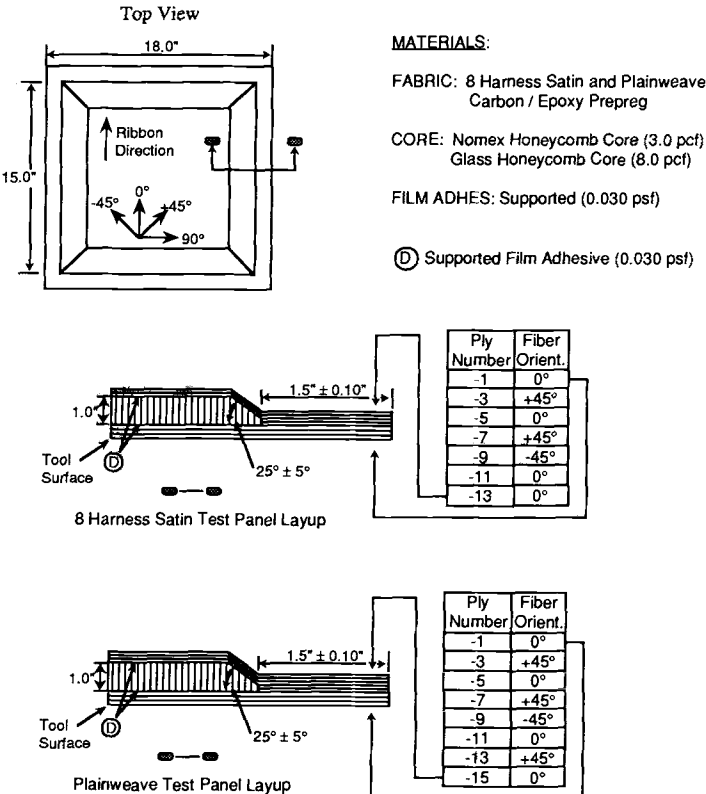
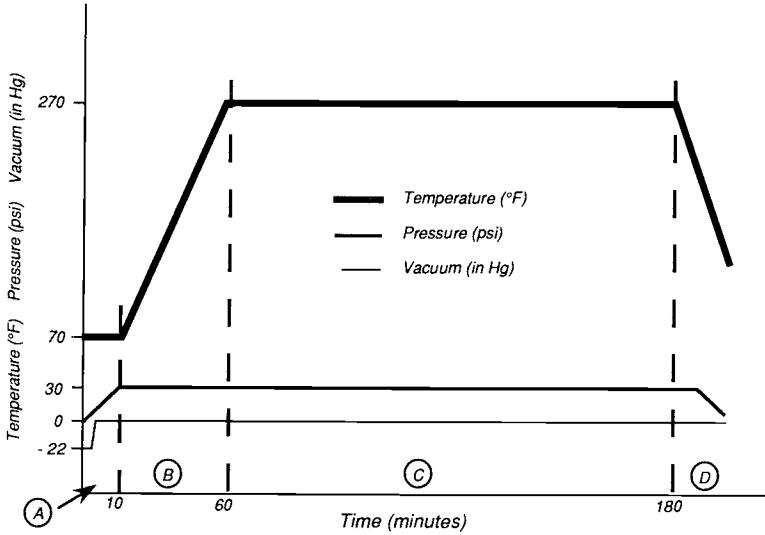
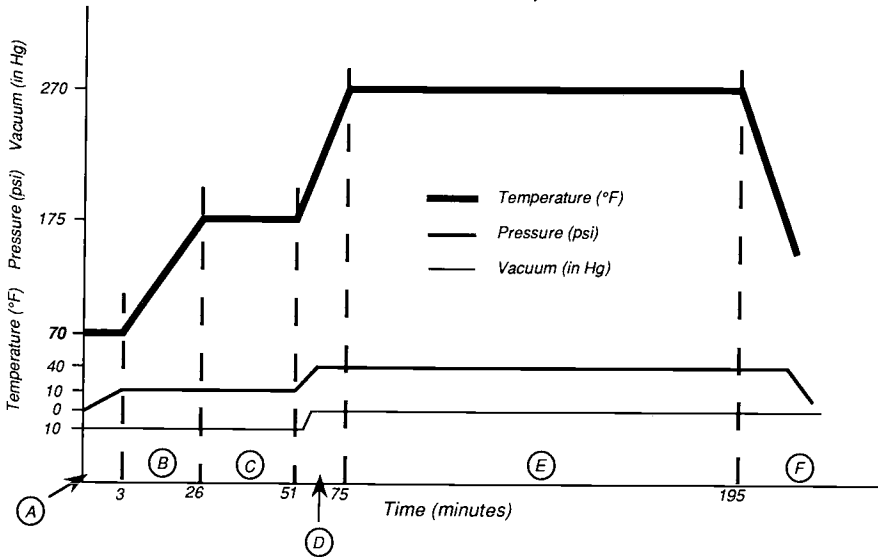


FIG. 2—Test panel lay-up configuration.



Segment A—Maintain 22 in. Hg (559 mm Hg) vacuum in bag. Increase vessel pressure to 30 psi (207 kPa). Vent vacuum when pressure reaches 10 psi (69 kPa). Segment B—Raise part temperature to $270 \pm 10^{\circ}\text{F}$ ($132 \pm 6^{\circ}\text{C}$) at a rate of 0.3 to $7.0^{\circ}\text{F}/\text{min}$ (0.2 to $4^{\circ}\text{C}/\text{min}$). Segment C—Hold at $270 \pm 10^{\circ}\text{F}$ ($132 \pm 6^{\circ}\text{C}$) for 120 min minimum. Hold begins when coldest TC reaches 260°F (127°C). Segment D—Cool part under pressure until part temperature reaches 140°F (60°C). Vent pressure.

FIG. 3—Baseline cure cycle.



Segment A—Maintain $5 \pm 5/-0$ in. Hg ($127 \pm 127/-0$ mm Hg) vacuum in bag. Increase vessel pressure to $10 \pm 5/-0$ psi ($69 \pm 34/-0$ kPa). Segment B—Raise part temperature to $175 \pm 10^{\circ}\text{F}$ ($79 \pm 6^{\circ}\text{C}$) at a rate of 0.3 to $7.0^{\circ}\text{F}/\text{min}$ (0.2 to $4^{\circ}\text{C}/\text{min}$). Segment C—Dwell at $175 \pm 10^{\circ}\text{F}$ ($79 \pm 6^{\circ}\text{C}$) for 25 min minimum. Dwell begins when coldest TC reaches 165°F (74°C). Segment D—Raise vessel pressure to 40 psi (276 kPa) minimum at a rate of 10 psi/min (69 kPa/min). Vent vacuum when pressure reaches 20 psi (138 kPa). Raise vessel temperature to $270 \pm 10^{\circ}\text{F}$ ($132 \pm 6^{\circ}\text{C}$) at a rate of 0.3 to $7.0^{\circ}\text{F}/\text{min}$ (0.2 to $4^{\circ}\text{C}/\text{min}$). Segment E—Hold at $270 \pm 10^{\circ}\text{F}$ ($132 \pm 6^{\circ}\text{C}$) for 120 min minimum. Hold begins when coldest TC reaches 260°F (127°C). Segment F—Cool part under pressure until part temperature reaches 140°F (60°C). Vent pressure.

FIG. 4—Step cure cycle.

Process and Material Variables

Processes

Baseline—The baseline process followed the production process for sandwich panels. The cure profile is shown in Fig. 3.

Step—The step process included an intermediate hold—a step—at $180^{\circ} \pm 10^{\circ}\text{F}$ ($82 \pm 6^{\circ}\text{C}$). The step was reached under reduced vacuum and reduced pressure. The step duration was timed to promote resin flow and to compact the laminates before the resin gelled. The maximum pressure allowed for sandwich panels was applied at the end of the hold, and the temperature was raised to the cure temperature. The cure profile is shown in Fig. 4.

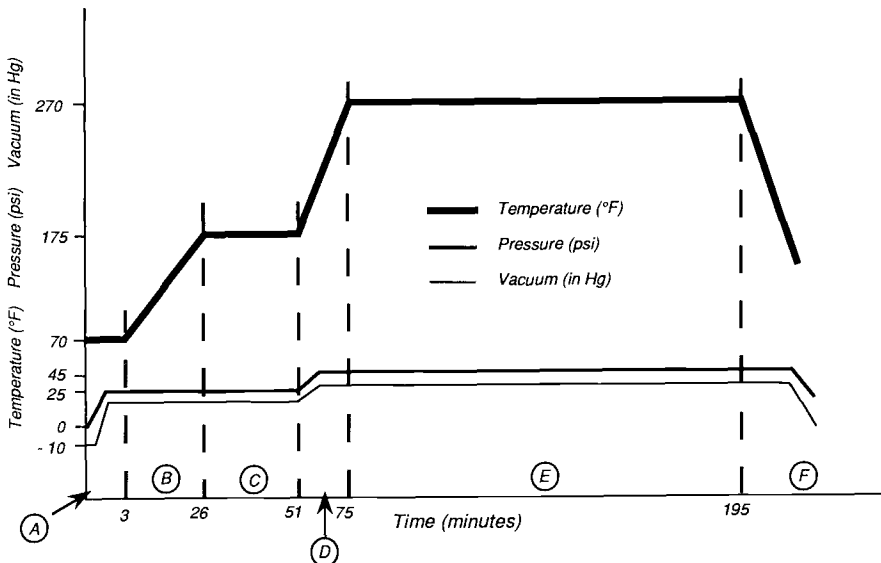
Internal Bag Pressurization (IBP)—IBP followed the step process. However, vacuum under the bag was replaced by positive pressure. The cure profile and the pressure schedule are shown in Fig. 5.

Baseline with Debulk—The plies of each facing were compacted under vacuum and at $150 \pm 10^{\circ}\text{F}$ ($66 \pm 6^{\circ}\text{C}$) during lamination. Panels were cured using the baseline cure profile.

Baseline with Precure—The facing on the tool side was laminated and cured separately. The assembly then continued over the cured skin. Panels were cured using the baseline cure profile.

Baseline with Tedlar®—A 0.002-in. (0.05 mm) ply of Tedlar was laminated on the bag surface. Panels were cured using the baseline cure profile.

Baseline with Surfacing Ply—A resin-rich surface supported with a thin carbon mat was laid down on the tool and bag sides. Panels were cured using the baseline cure profile.



Segment A—Maintain $5 \pm 5/-0$ in. Hg ($127 \pm 127/-0$ mm Hg) vacuum in bag. Increase vessel pressure to 25 psi (172 kPa). When vessel pressure reaches 10 psi (69 kPa), slowly bleed off vacuum and run in 15-psi (103 kPa) pressure under the bag. **Segment B**—Raise part temperature to $175 \pm 10^{\circ}\text{F}$ ($79 \pm 6^{\circ}\text{C}$) at a rate of 0.3 to $7.0^{\circ}\text{F}/\text{min}$ (0.2 to $4^{\circ}\text{C}/\text{min}$). **Segment C**—Dwell at $175 \pm 10^{\circ}\text{F}$ ($79 \pm 6^{\circ}\text{C}$) for 25 min minimum. Dwell begins when coldest TC reaches 165°F (74°C). **Segment D**—Raise vessel pressure to 45 psi (310 kPa) at a rate of 10 psi/min (69 kPa/min). Increase pressure under the bag to 30 psi (207 kPa). Raise vessel temperature to $270 \pm 10^{\circ}\text{F}$ ($132 \pm 6^{\circ}\text{C}$) at a rate of 0.3 to $7.0^{\circ}\text{F}/\text{min}$ (0.2 to $4^{\circ}\text{C}/\text{min}$). **Segment E**—Hold at $270 \pm 10^{\circ}\text{F}$ ($132 \pm 6^{\circ}\text{C}$) for 120 min minimum. Hold begins when coldest TC reaches 260°F (127°C). **Segment F**—Cool part under pressure until part temperature reaches 140°F (60°C). Vent pressure.

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FIG. 5. Internal bag pressurization (IBP) cure cycle.

Materials

Fabric—Two fabric styles were included, an eight-harness satin-weave carbon fabric with an areal weight of 360 g/m², and a plain-weave carbon fabric with an areal weight of 190 g/m². These fabric styles are the dominant material types used in production panels.

Resin—The resin from one qualified supplier was used. This resin is Hexcel F522, a 250°F (121°C) curing, self-adhesive, epoxy resin.

Core—Two core types were used, a 3-lb/ft³ (48-kg/m³) density Nomex core with 0.125-in. (3-mm) cells, and an 8-lb/ft³ (128-kg/m³) density glass core with 0.188-in. (5-mm) cells.

Resin Content—The fabrics were impregnated at two levels, 39 and 45%/w.

Test Procedures

Leak Testing

Leak testing was performed by submerging the panel in a water bath maintained at 170 ± 10°F (77 ± 6°C). Air entrained was first brushed off the surfaces before the observations would begin. As the panel warmed, the pressure of the air trapped in the honeycomb cells would rise, forcing the air through the void network to form bubbles or an air stream on the surface of the panel at a leak site. A panel would fail the leak test if an air bubble formed within 2 min of immersion.

Compression Testing

Compression specimens were cut from the bag side and the tool side of each panel (Fig. 6). The specimens were machined and tabbed (Fig. 7), then tested at room temperature per the

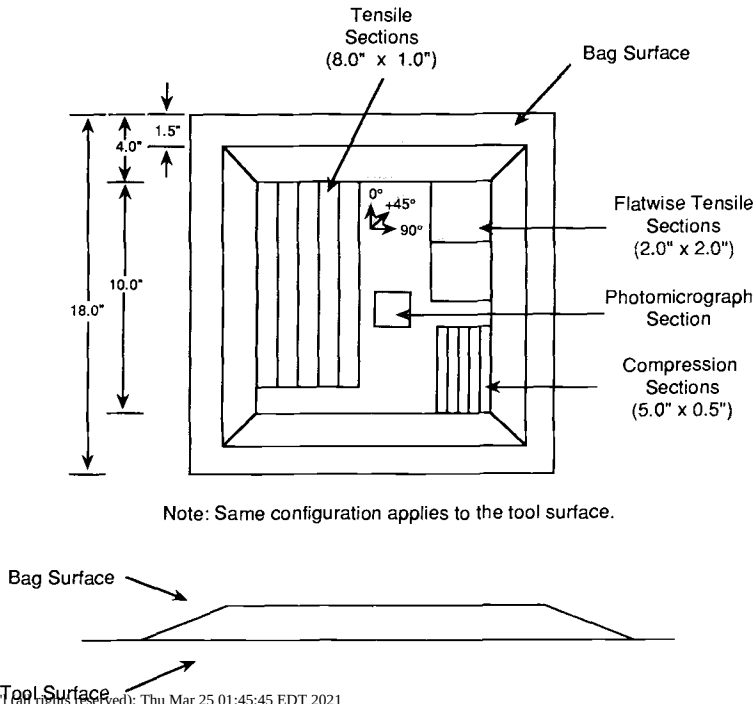


FIG. 6—Test panel sampling configuration.

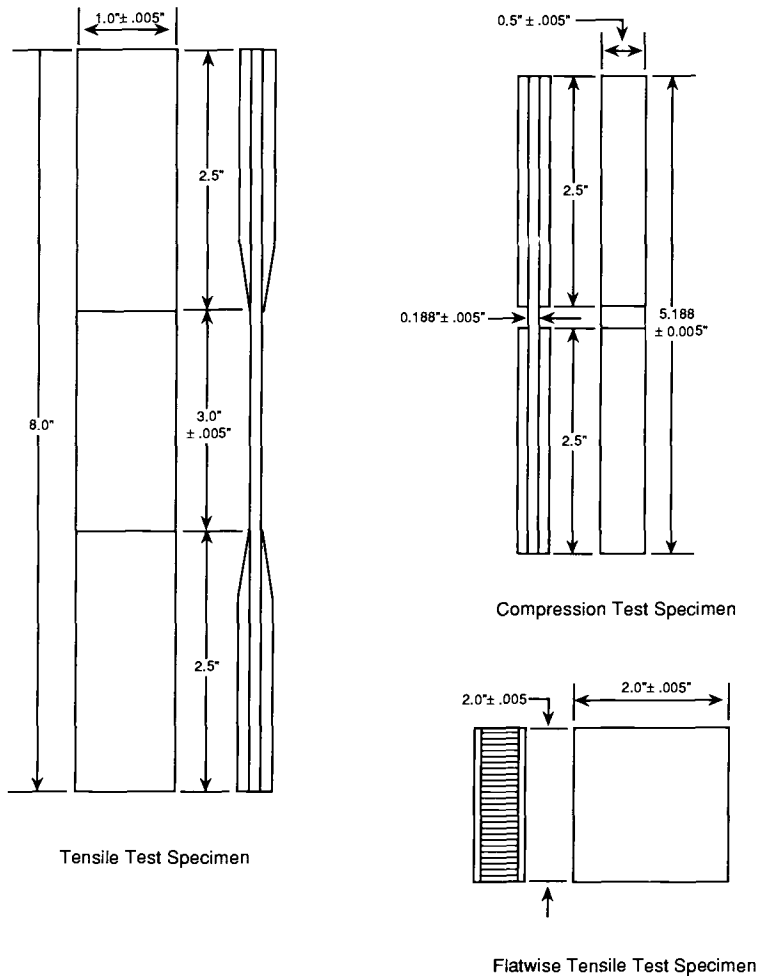


FIG. 7—Test panel mechanical testing specimen configuration.

ASTM Test for Compressive Properties of Unidirectional or Crossply Fiber-Resin Composites (D3410), using a crosshead speed of 0.05 in./min (1.3 mm/min).

Tension Testing

Tension specimens were cut from the bag side and the tool side of each panel (Fig. 6). The specimens were machined and tabbed (Fig. 7), then tested at room temperature per the ASTM Test for Tensile Properties of Fiber-Resin Composites (D3039), using a crosshead speed of 0.05 in./min (1.3 mm/min).

Flatwise Tension Testing

Flatwise tension specimens were cut from each panel (Fig. 6) and bonded to aluminum blocks (Fig. 7). The specimens were tested at room temperature per the ASTM Test for Tension Test of Flat Sandwich Constructions in Flatwise Plane (C297), using a crosshead speed of

0.05 in./min (1.3 mm/min).

Test Results

Screening

The screening plan is summarized in Table 1. One panel for each condition shown was made except for the step cycle and for the IBP cycle where two additional panels for each cycle were fabricated to confirm leak and compaction results. Each panel was leak tested, assessed for surface finish, then cross sectioned. A skin compaction index was calculated for each facing as follows.

C_i = 1000*(t_{max} - t_{min})/t_{nom}

where

- C_i = compaction index, (in /in)*1000,
- t_{max} = maximum skin thickness, in ,
- t_{min} = minimum skin thickness, in., and
- t_{nom} = nominal skin thickness, in

Table 2 shows the surface finish assessment, and Table 3 shows the leak test results. As anticipated, precuring the skin or adding a Tedlar film does seal the facing affected. The other treatments, surfacing ply, high resin content prepreg, and debulk, bring no relief. The two process modifications, step and IBP, result in better panels overall. The compaction results confirm this assessment (Figs 8 and 9), with the standard process giving the worst results, followed by the step process and the IBP process. Figure 8 also shows the effect of precuring and of the hot debulk. The precured skin results in the best compacted laminate on the tool side, debulking improves the quality of the laminate on both sides.

Compaction and laminate quality are correlated as seen in the photomicrographs (Figs 10 a through f). The standard process is characterized by severe ply separation between the core walls, fiber misalignment, and voids unfilled with resin (Fig 10a and b). The step process is characterized by less ply separation, fewer voids, and interlaminar cavities filled with resin (Fig 10c and d). The IBP process results in better laminates (Fig 10e and f).

Mechanical Testing

Following screening, the better panels—step and IBP—were retained for mechanical testing and compared to the baseline process. Tension, compression, and flatwise tension results are shown in Tables 4 through 6.

TABLE 1—Process variable screening matrix

Core Density, lb/ft ^{3a}	Resin Content (%) ^b	Baseline Cure Cycle		Step Cure Cycle		IBP Cure Cycle	Graphite Surfacing Ply	Tedlar Surfacing Ply	Hot Debulked Tool Facing	Precured Tool Facing
		PW	8HS	PW	8HS	8HS	8HS	8HS	8HS	8HS
3.0	39.0	1	1	1	2	2	1	1	1	1
3.0	45.0		1							
8.0	39.0	1	1	1	2	2				
8.0	45.0		1							

TABLE 2—Surface finish evaluation ^a

Core Density, lb/ft ^{3b}	Resin Content, % ^c	Baseline Cure Cycle		Step Cure Cycle		IBP Cure Cycle	Graphite Surfacing Ply	Tedlar Surfacing Ply	Hot Debulk Tool Facing	Precured Tool Facing
		PW	8HS	PW	8HS	8HS	8HS	8HS	8HS	8HS
3 0	39 0	7/5 ^d	3/5	6/1	2/1	1/1	2/2	3/2	2/5	2/5
3 0	45 0		3/5							
8 0	39 0	7/1	3/1	6/1	2/1	1/1				
8 0	45 0		3/1							

^a Surface finish criteria 1 best and 7 worst

1 Surface uniformly resin filled

2 Completely filled surface with presence of surfacing ply

3 Resin-filled surface with randomly scattered minor porosity

4 Resin-filled surface with randomly scattered severe porosity

5 Resin-filled surface with areas of resin starvation

6 Resin-filled surface with minor mudcracking

7 Resin-filled surface with severe mudcracking

^b 1 lb/ft³ = 16 kg/m³

^c Resin content ± 3 0% tolerance

^d Tool facing → x/y ← bag facing

The flatwise tension test is not a good discriminator because all failures occurred away from the skins

The tension and compression test results show that the process and material variables had a significant effect on the facing strength The IBP process results in the better tensile and compressive properties The plain-weave fabric results in significantly greater tensile and compressive properties Core density also affects tensile results, with the 8-lb/ft³ (48-kg/m³) core resulting in improved tensile strength.

Discussion

The results of the small panel test led to the use of the step cycle in production In general, these results apply to the production panels the surface finish becomes acceptable and the

TABLE 3—Leak testing results

Core Density, lb/ft ^{3a}	Resin Content, % ^b	Baseline Cure Cycle		Step Cure Cycle		IBP Cure Cycle	Graphite Surfacing Ply	Tedlar Surfacing Ply	Hot Debulk Tool Facing	Precured Tool Facing
		PW	8HS	PW	8HS	8HS	8HS	8HS	8HS	8HS
3 0	39 0	∅ ^c	●	∅	∅	●	●	∅	●	∅
3 0	45 0		●							
8 0	39 0	∅	●	∅	∅	∅				
8 0	45 0		●							

^a 1 lb/ft³ = 16 kg/m³

^b Resin content ± 3 0% tolerance

^c Tool facing → ∅ ← bag facing Filled half-circle indicates leak(s) evident on that facing

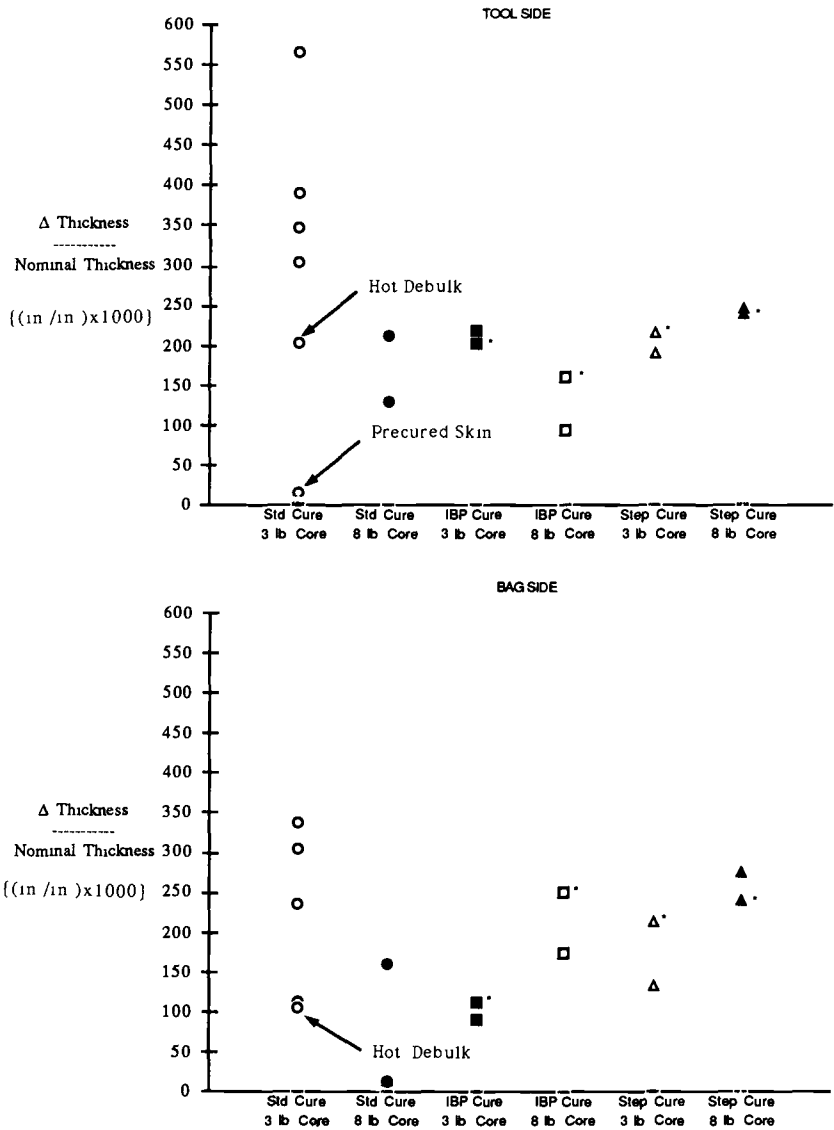


FIG 8—Eight-harness satin-weave compaction index results (* means data point from additional test panels)

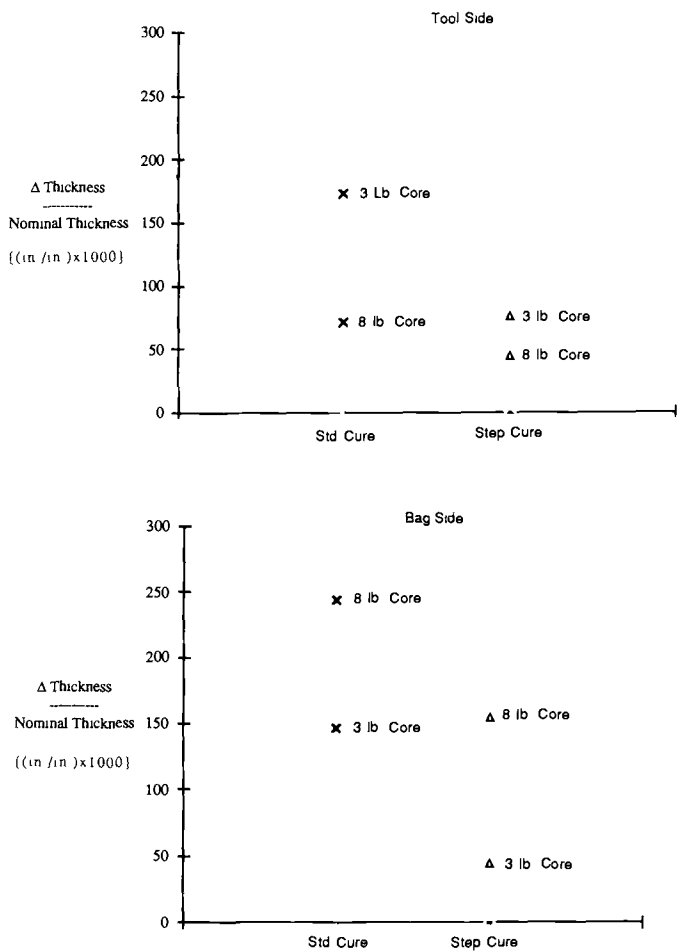


FIG 9—Plain-weave compaction index results

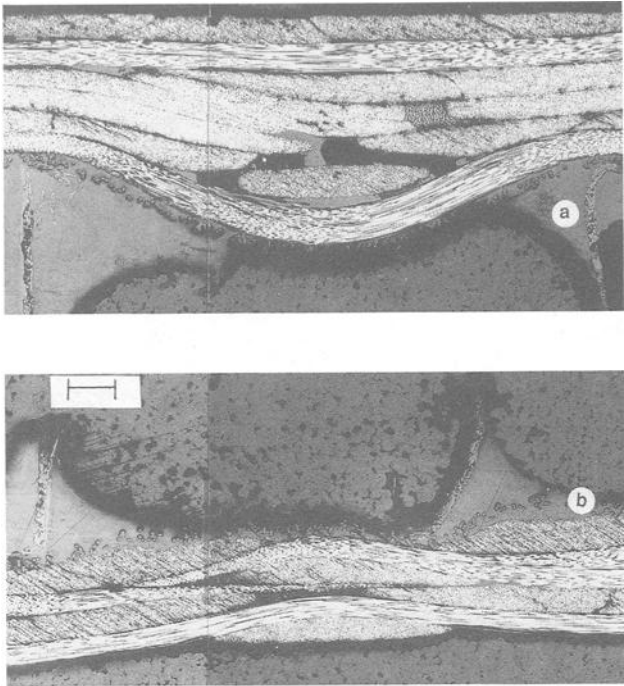


FIG. 10—Typical laminate cross section cured with the baseline cure cycle: (a) tool surface and (b) bag surface. Scale marker indicates 250 μm .

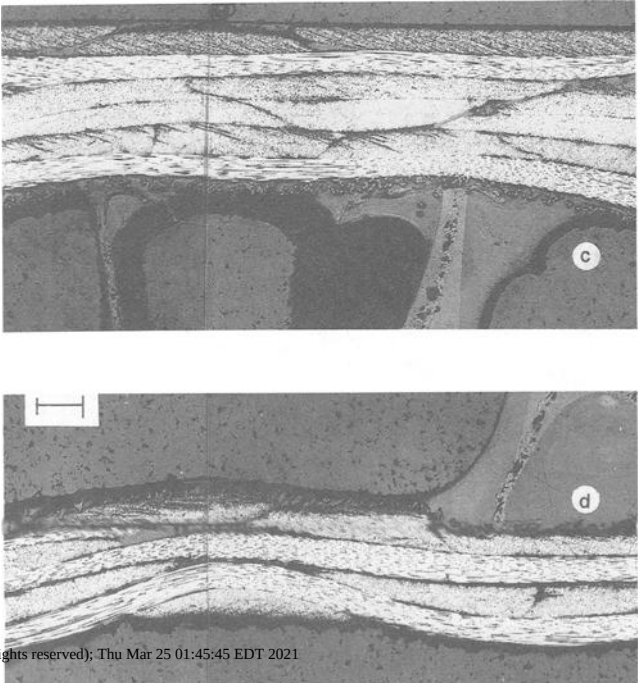


FIG. 10—Typical laminate cross section cured with the step cure cycle: (c) tool surface and (d) bag surface. Scale marker indicates 250 μm .

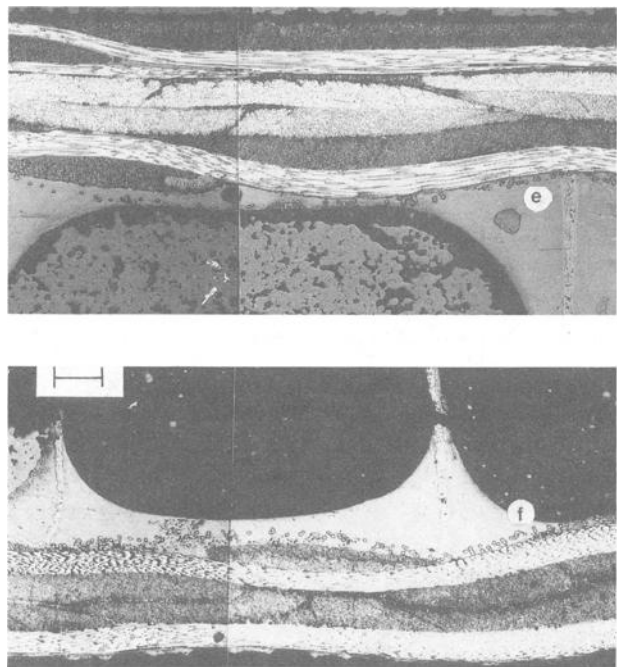


FIG. 10—Typical laminate cross section cured with the IBP cure cycle: (e) tool surface and (f) bag surface. Surface marker indicates 250 μm .

TABLE 4—Flatwise tension testing results in ksi.^a

Fabric	3-lb Honeycomb Core, ^b psi	8-lb Honeycomb Core, ^c psi	Cure Cycle
8HS	360	1266	baseline
PW	321	1000	
8HS	335	1196	
PW	322	1100	step
8HS	324	1139	
PW	M ^d	M	IBP

^a Ultimate stress results shown. 1 psi = 6.89 kPa.
^b All 3-lb (1.4-kg) honeycomb core coupons exhibited 100% core failure.
^c All 8-lb (3.6-kg) core coupons exhibited 100% core-to-facing bond failure.
^d M: missing data

TABLE 5—Tension and compression testing stress results in ksi (standard deviation in parentheses) ^a

Fabric	3-lb Honeycomb Core ^b				8-lb Honeycomb Core ^b				Cure Cycle
	Bag Facing		Tool Facing		Bag Facing		Tool Facing		
	Tension	Comp	Tension	Comp	Tension	Comp	Tension	Comp	
8HS	74.0 (4.6)	34.8 (4.4)	68.3 (3.4)	35.2 (2.3)	80.3 (7.4)	32.6 (7.8)	68.1 (9.3)	44.1 (5.0)	baseline
PW	69.0 (5.1)	42.9 (4.3)	81.3 (5.2)	48.8 (3.0)	82.1 (3.8)	50.4 (4.9)	88.8 (3.8)	55.4 (8.6)	
8HS	86.5 (14.4)	38.1 (9.7)	62.8 (7.3)	38.0 (6.7)	103.7 (4.4)	48.3 (10.9)	74.5 (8.7)	48.2 (6.4)	step
PW	80.9 (3.9)	48.7 (3.3)	89.1 (4.7)	50.9 (1.8)	83.1 (3.1)	49.4 (4.2)	79.9 (4.9)	50.0 (5.6)	
8HS	71.7 (4.6)	30.9 (9.5)	70.7 (5.9)	39.5 (4.9)	101.9 (3.3)	57.7 (2.8)	82.8 (5.2)	53.9 (3.6)	IBP
PW	M ^c	M	M	M	M	M	M	M	

^a Ultimate strength results indicated 1 ksi = 6.89 kPa

^b 1 lb = 0.5 kg

^c M missing data

incidence of leaks is reduced (nonexistent for the panels made with plain-weave fabrics). The small panel test is now in general use and has been particularly helpful in discriminating among competing resin systems.

Conclusions

1 The two process modifications tried—step and IBP—result in sandwich panels with better sealing, better compaction, fewer interlaminar voids, better surface finish, and equal or better strength than the baseline panels.

2 The step cycle is considered superior to the IBP cycle because it results in consistently sealed panels. However, a change in the IBP pressure schedule may alter this conclusion.

TABLE 6—Tension testing outer fiber strain results in $\mu\text{in/in}$ (standard deviation in parentheses) ^a

Fabric	3-lb Honeycomb Core ^b		8-lb Honeycomb Core ^b		Cure Cycle
	Bag Facing	Tool Facing	Bag Facing	Tool Facing	
8HS PW	9 546 (1581) 10 881 (829)	10 153 (596) 12 162 (180)	11 139 (1663) 11 667 (690)	10 341 (1674) 13 302 (463)	baseline
8HS PW	9 911 (1960) 11 939 (783)	9 078 (1138) 12 518 (621)	11 758 (449) 12 146 (986)	10 713 (1285) 11 423 (854)	step
8HS PW	9 500 (462) M	10 092 (975) M	12 162 (739) M	12 621 (1571) M	IBP

^a Ultimate strain results shown as measured with strain gages 1 $\mu\text{in/in}$ = 1 $\mu\text{m/m}$

^b 1 lb = 0.5 kg

^c M missing data

3 Other material changes—Tedlar and precured skin—are only considered as stand-by solutions. The Tedlar seals the panel without reducing the interlaminar problems. The precured skin remains undesirable because of the cost involved in this solution.

4 The use of a surfacing ply results in a good surface finish but an unsealed panel.

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- [1] Brayden, T. H. and Darrow D. C., "Effect of Cure Cycle Parameters on 350°F Cocured Epoxy Honeycomb Core Panels," in *Proceedings of the 34th International SAMPE Symposium*, 1989, pp. 861–874.
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An Evaluation of the Current Status of Automated Process Control for Thermosetting Composites

REFERENCE. Holl, M. W. and Rehfield, L. W., "An Evaluation of the Current Status of Automated Process Control for Thermosetting Composites," *Composite Materials: Testing and Design (Tenth Volume)*, ASTM STP 1120, Glenn C. Grimes, Ed., American Society for Testing and Materials, Philadelphia, 1992, pp. 308–319.

ABSTRACT. Automated process control for thermosetting composite cure is used to ensure part quality while reducing the energy required to manufacture parts. This study summarizes the techniques used in automated process control and evaluates the progress toward implementing these methods in production. An ideal transducer must provide information about the physical state of the resin while the part is in the processing tool. It must also be able to survive the hostile environment of processing equipment, be simple to install, not interfere with the mechanical properties if it is left behind in the part, and, of course, be economical. Promising transducers based on dielectric, acoustic, and fluorescent signals are detailed in this study. When the physical state is known, the control algorithm acts on the information to vary the cure parameters (temperature, pressure). A number of different strategies for controlling the cure have been proposed. All are developed by trial and error and are part-specific. As more is understood about the physical mechanisms of resins during processing, algorithm development will be easier.

KEYWORDS. composite materials, manufacturing, cure, cure monitoring, process control

Much research and development of automated process control (APC) was done in the 1980s. Without APC, parameters of the cure schedule (temperature, pressure) are preset and inflexible. Differences in the prepreg material from part to part do not alter the cure because no variation is allowed. The cure schedule represents what will work most of the time, rather than the most efficient for each individual part.

APC acquires physical information about the resin while it is being processed and subsequently controls the cure parameters (Fig. 1). The cure schedule "adapts" to each part and provides the most efficient cure (shortest while maintaining quality).

Two areas of APC that required the most new thinking were in the development of a transducer and a set of control laws, here termed the *algorithm*. The transducer has to be able to withstand the hostile environment of composite processing equipment while measuring some physical quantity related to the viscosity and degree of cure of the resin. Transducers used in APC include microdielectric-, fluorescence-, and acoustic-based sensors. The most experience is with microdielectric sensors. Research and development continues on all of these, and production implementation of a microdielectric-based APC system is in progress.

¹ Graduate research assistant and professor and division leader for Aeronautical Engineering, respectively, University of California, Department of Mechanical, Aeronautical, and Materials Engineering, Davis, CA 95616.

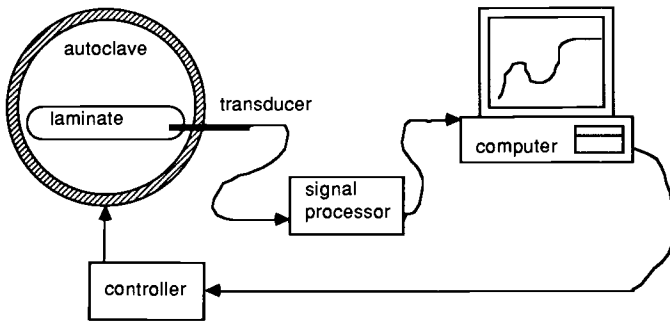


FIG. 1—Automated process control system.

The algorithm must be able to use the knowledge of the resin's present and past physical state for setting future cure events and parameters. Algorithms are developed by trial and error. This is expensive and leads to the algorithm being specific to a certain part geometry.

Micromet and Applied Polymer Technology, Inc. (APT) are developing a total cure monitoring and control package for research and production. Their goal is to produce a low-cost turn-key system with a low learning curve. In this way, the high costs of development and implementation can be minimized. General Dynamics/Pomona Division has developed an APC system for a production line. This system is ready for implementation for making a polyimide part.

Transducers

Dielectric Transducers

The advantage of monitoring the cure with dielectric measurements is that the process can be monitored as it goes through all states of the material (that is, liquid to solid) with one transducer. APC was helped by the invention of the microdielectric sensor. Before the introduction of the microdielectric sensor, dielectric measurements were taken with parallel plate electrodes.

In the parallel plate type of measurement, things like thermal expansion or resin shrink can change the distance between plates. The dielectric signal is a function of this distance between plates, and there is obvious difficulty in continuously measuring this distance during cure. Microdielectric sensors solved this problem by incorporating tiny electrodes on the surface of a probe. The spacing between electrodes remains fixed making them a viable sensor for APC.

One end of the thin, ribbonlike microdielectric transducer (Fig. 2) is placed in contact with the prepreg in the vacuum bag during the layup of the part. The other end exits out of the vacuum bag and is connected to the data acquisition equipment. The end inside the bag contains the sensor, which is wet out with resin during cure. An alternating current is applied to the electrodes, and the induced signal is read by equipment from which the dielectric information can be interpreted.

In a production environment, technicians are responsible for installing the transducer during the layup of a part. This can be time-consuming and cumbersome [1]. The transducer going through the bag creates a spot for vacuum leaks during cure. Also, the connecting wires must be inspected to ensure they are standing up to the thermal cycling of the processing equipment and can still function. These details limit the attractiveness of microdielectric

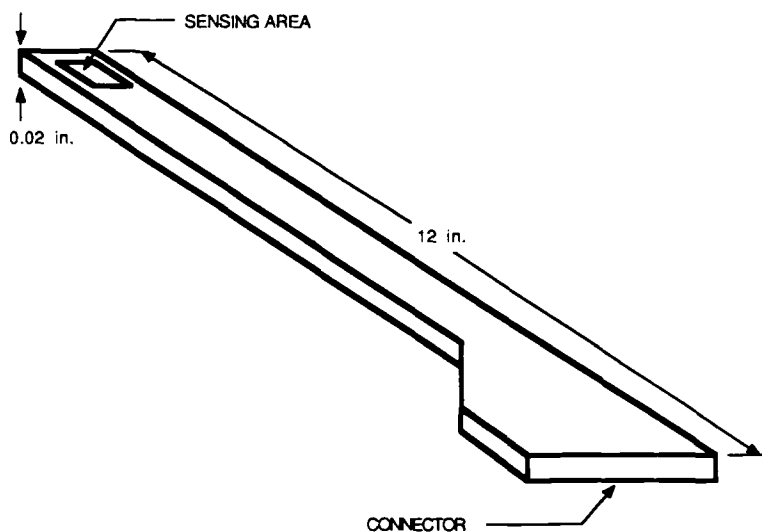


FIG. 2—Schematic of microdielectric transducer.

If conducting fibers are allowed to touch the electrodes of the sensor and short them out, there is no signal and no intelligent control of the cure process. This drawback of the microdielectric technique was solved by covering the sensor with a permeable membrane. Resin can flow in while the miniature electrodes are shielded from the fibers of the prepreg.

If the sensor is placed between plies, it becomes a permanent part of the finished laminate. If the sensor is placed in the bleeder plies, no information is available until the resin flows to the sensor, but it does not become an integral part of the laminate. These trade-offs must be considered when developing an APC system. General Dynamics/Pomona Division designed a tab into the part that would contain the sensor during the cure and be machined off afterwards. This technique required extra development to ensure that monitoring cure at the edge of a part would predict cure in other locations [2].

Early models of microdielectric sensors could not survive the rugged handling of the production floor and the extreme environment of process tooling. This problem has been solved by new sensors developed for these conditions. For example, the IDEX® sensor from Micromet functions in the -238 to 707°F (114.4 to 375°C) range and has no pressure limit.² Although no data on the failure rate of microdielectric sensors are available, it is the consensus of people working with them that failure is a possibility but no longer a major problem.

Ions and dipoles are present in most resins. If electrodes are put into a resin and given a potential difference, the ions tend to "travel" toward the appropriate electrode. The dipoles on the resin molecules tend to align themselves in the electric field. During thermosetting cure, bonds form that link molecules together (cross-linking). This affects properties such as viscosity, the degree of cure, and the ability of the ions to move and the dipoles to align themselves.

As an alternating current is applied to the electrodes on the microdielectric sensor, the ions and the dipoles react to the electric field. The conductance (G) and the capacitance (C) are measured, but these quantities are related to size of the sample. More useful to APC are the quantities of conductivity (σ) and dielectric relaxation (τ), which can be derived from C and

G and are invariant with respect to the size of the sample. The conductivity (σ) is the value often correlated with viscosity and degree of cure, which are the physical values important to APC [3,4]

In a review of dielectric cure monitoring, Ciriscioli and Springer presented the Kranbuehl method for obtaining σ from G and C [5] The equation

$$\epsilon_m''\omega \approx \sigma/\epsilon_o = \text{constant} \quad (1)$$

where

ϵ_m'' = the measured loss factor,
 ω = the angular frequency, and
 ϵ_o = the permittivity of free space,

is valid only at frequencies where the product $\epsilon_m''\omega$ is constant

$$(\epsilon_m''\omega)_1 = (\epsilon_m''\omega)_2 = \dots = \text{constant} \quad (2)$$

Frequencies that do not meet this requirement are not valid for computing σ in Eq 1

These equations demonstrate that the frequency dependence of the dielectric signal has been accounted for. Because the correct frequencies are not known before starting the cure, the equipment must take readings over a wide range of frequencies and "find" at least two that satisfy Eq 2. In practice, it takes about 2 min to scan the range of frequencies and apply Eq 1 to get a σ value. Time of measurement has not been a limiting factor in developing dielectric APC techniques.³

When the resin system is simultaneously subjected to a low frequency at a very low viscosity, the ions are mobile enough and have enough time to accumulate around the electrodes. This forms a capacitive boundary layer around the electrodes and affects the measurements. At higher frequencies and low viscosities, this phenomenon, known as electrode polarization, is not observed because the mobile ions do not have time to travel and collect around the electrodes [6]. Electrode polarization can affect the true value of σ and must be accounted for. Demonstrations of the accountability are in the literature [5,6].

The ionic concentration of prepreg material may vary from part to part, but this does not affect the accuracy of the measurement. Even if the ionic concentration changes a small percentage, the ionic mobility changes decades in magnitude (10^{-10} to 10^{-5}). Changes in ionic concentration are negligible in comparison.³

The similarity in the shapes of the viscosity and conductivity versus time graphs demonstrates their interdependence. The graphs are not identical, but the conductivity follows the viscosity closely enough to make intelligent decisions in APC. This has been demonstrated in the literature [2-4].

The conductivity value can be thought of as a relative measure of the ionic impurities' ability to travel through the resin. As the viscosity decreases in the resin, the ability of ion travel goes up, as the degree of cure gets larger, the ability of ion travel goes down. The conductivity depends on certain states of the resin, namely viscosity and degree of cure. Conductivity has been empirically linked to viscosity, but it must be remembered that conductivity is what is being measured. It is not important to link precisely the viscosity to conductivity if, by knowing an interdependence exists, a working closed loop cure control method can be developed based on the highly accurate measurement of conductivity.

³ D. Kranbuehl, Department of Chemistry, College of William and Mary, Williamsburg, VA, personal

Fluorescence-Based Transducers

The fluorescence of a resin has been found to depend on its viscosity [7]. To exploit this interdependence as a cure control method, excitation energy must be delivered to a resin and the emitted fluorescence must be carried back to the data acquisition equipment. The most promising way to take these measurements is demonstrated by the fluorescence optrode cure sensor (FOCS) developed at McDonnell Douglas Research Laboratories in 1983–1985 [8]. A simplified diagram of the sensor as it was used in the development appears in Fig. 3.

FOCS incorporates fiber optics to allow the light to travel in and out of the extreme auto-clave environment. The fiber optic method also provides immunity to electromagnetic interference, can easily be multiplexed, and has a potential for low-cost applications [8].

Resin from the curing laminate would flow through tiny flow holes in the tool to a resin cavity under the surface of the tool. In the resin cavity, the resin is exposed to light brought in by the optical fiber. The resulting fluorescence is also carried back through the same fiber.

Because the resin cavity is lined with Teflon® release film, the old resin in the cavity is easily removed from the tool after cure. When the Teflon film is replaced and the stainless steel flow plate is reinstalled, the tool is ready to be used again. The transducer is not in the way when technicians are laying up a part because it sits under the surface of the tool. Also, there are no reasonable temperature and pressure limitations. The FOCS has no trouble performing in the hostile environment of presses and autoclaves.

The above advantages lead to the reduction of costs associated with APC. The material cost per cure cycle is extremely low because the only disposable object involved is the release film. The labor installation costs are low because the only work involved after mounting the sensor on the tool is the replacement of the release film in the resin cavity. Because the sensor stays out of the way, layup is much less cumbersome than with sensors that must be embedded in the prepreg.

One negative aspect of this transducer is that only one side of the laminate is sampled. Some algorithms being considered for APC, to be discussed subsequently, require information at the surface and the middle of the laminates. These algorithms have been developed with dielectric cure sensors in mind, however. As the fluorescence technology becomes more mature, it can be expected that algorithms tailored around its strengths will be developed as well.

One alternative is using mathematical cure models to translate properties at the surface to

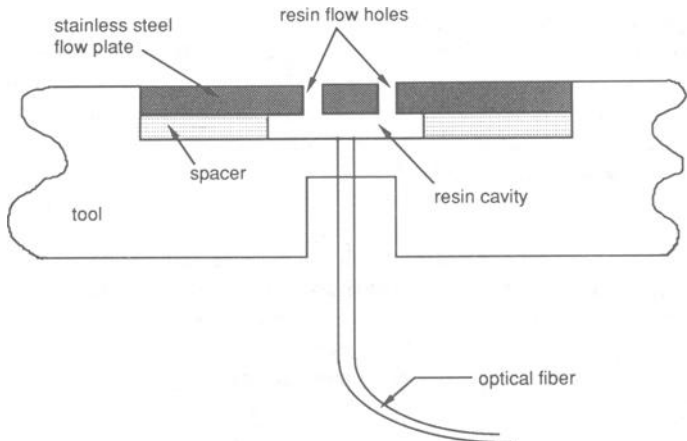


FIG. 3—Schematic of interface between resin and optical fiber for FOCS sensor.

properties at other parts of the laminate. The drawbacks of using only models as a means of control have been outlined in Ref 2, but the use of a sensor with a model may alleviate some of these problems.

In a low-viscosity resin, when the molecules go from the excited state to the nonexcited state, they can lose their energy by intramolecular torsional motions. There is no radiation released. When this torsional motion is restricted by a higher viscosity, the molecules lose the energy by fluorescing [7]. In this way, the link is established between viscosity and the fluorescent signal.

The intensity and the maximum wavelength of the signal have been used to monitor the viscosity and the degree of cure. The time profiles of these quantities (shapes of the curves with respect to time) are indeed similar, but no example of an APC system using the fluorescence technique is available yet.

According to the study conducted by the Air Force Wright Aeronautical Laboratories (AFWAL) [8], the next stages in the development of the sensor will be the control of the sensor response variations that inhibit signal reproducibility. However promising to APC, work remains on characterizing the signal of the FOCS. Once this is completed, it can begin to be incorporated into an APC system.

Acoustic-Based Transducers

Because acoustic waves travel by molecular interaction they are sensitive to physical changes in the medium through which they are passed. The physical changes of interest in this case are, of course, viscosity and degree of cure of the resin. With this in mind, attempts have been made to develop an acoustic-based transducer that can be used for real-time control.

Conventional techniques have a transmitter and receiver on either side of the laminate (Fig. 4). The resulting signal is a measure of the energy absorbed by the material. This technique does not differentiate between points through the thickness of the laminate and depends on the amount of material [9]. This leads to a dependence on the thickness of the laminate. When thick laminates are monitored, the attenuation of the signal is very large.

An approach to this problem has been developed by R. T. Harrold at Westinghouse [10]; small fibers have been buried in composite layups to act as guides for the acoustic waves (Fig. 5). They were appropriately named acoustic wave guides (AWG). Energy escapes from the AWG relative to the acoustic impedance between the wave guide and the resin. There is no dependence on the amount of material. The quantities measured are the sonic output, a measure of the magnitude of the signal that gets through the wave guide, and the transit time, a measure of the speed of the signal. It is hoped that the use of a wave guide will be less expensive and perform better than through-the-thickness techniques.

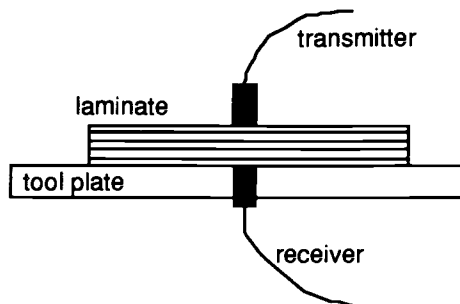


FIG. 4 Through-the-thickness acoustic transducer.

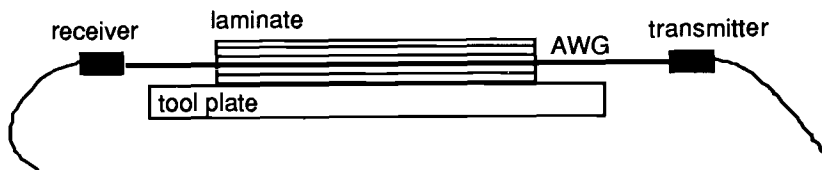


FIG. 5—Acoustic wave guide transducer.

The effects of process variables on the AWG signal are not fully understood [8]. It is common to use different fiber orientations in composite parts, and varying the fiber direction produced a variation in the AWG signal. Changing resin pressure also contributed to a lack of signal reproducibility. The coupling mechanism between the wave guide and acoustic transmitter and receiver was not sufficient to provide a complete bond every time when Harrold and coworkers did the original research. This contributed to signal attenuation and a lack of reproducibility, but a suitable mechanism would seem a minor problem to overcome. All these effects on the signal must be clearly understood before the AWG signal is useful as a production APC sensor.

In both the conventional and wave guide methods, the equipment must be made of material able to survive the temperatures and pressures of production processing equipment. Conventional through-the-thickness acoustic transducers are acceptable. For the AWG, nichrome (nickel-chromium) has been identified as a material able to withstand the conditions and still be sensitive enough to get valuable readings [8]. More research may produce advances on the wave guide material.

The wave guide is embedded in the prepreg layup and becomes an integral part of the finished part. The wave guide can be as small as 0.06 in. (0.15 cm) in diameter. The disadvantage of leaving it behind in the finished part is small in comparison to the dielectric transducer. The acoustic wave guide also has potential as a device to monitor stresses, strains, and other material properties later in the life of the part [11].

The bonding of transducer to material in both conventional transducers and the AWG is important because a complete bond has a different effect on the signal than a partial bond. An incomplete bond was a source of variation in the AWG signal [8]. Further development of some form of coupling mechanism would eliminate this problem for both methods. It may also add to the ease of use in a factory and make it easier on production people responsible for making the connection.

Effects of Voids

Void-producing gases are by-products of the thermosetting chemical reactions of some resins systems; polyimides provide a popular example. Voids may also exist when the resin system doesn't produce any gaseous by-products (for example, some epoxies as a result of entrapped moisture and other impurities in the prepreg during the cure) [12]. The voids are always undesirable because they reduce the mechanical properties of the laminate, but their existence to some degree is unavoidable. Their effects must be accounted for on the performance of the transducer.

In reviewing the physical basis for each of the transducers discussed, it is evident that voids in contact with the sensor have some effect. The critical void concentration in which impact on the transducer signal is no longer negligible is an important quantity to be determined, but a quantitative analysis goes beyond the scope of this investigation. These critical concentrations are not available in the open literature, if known at all. Besides knowing when voids

become important, there should also be a way to continue monitoring with the transducer if they appear

An example of the need to account for the effects of volatiles (gases being produced and evolved during the cure) on a microdielectric signal is given by the APC system being implemented at General Dynamics/Pomona Division ⁴ Both a change in the viscosity and a change in the evolution of volatiles had an effect on the conductivity signal, and the contributions from both were determined experimentally. They were then separated to analyze the viscosity effects, which were of primary importance ⁴

Algorithms

If an infallible transducer existed that would give exact profiles of resin viscosity and degree of cure in time, the question of when to apply temperature and pressure, and at what magnitude, would still remain. The set of instructions that interprets information from a transducer and acts to control the cure process is called the "algorithm."

Strategies to control cure have been proposed and are discussed below. To develop an algorithm using these strategies, it would be necessary to go through many processing cycles with an actual or scaled down part. The time and magnitude of cure variable application can be systematically varied and evaluated for part quality. Once the optimums are established, the points of application can be correlated to the transducer signal. In the working APC system, the algorithm "watches" for the appropriate signal from the transducer to trigger and control the cure parameters.

The reactant ratios of the resin in prepreg material can vary from lot to lot. Also, the prepreg's history of exposure to humidity and room temperature can be different [1]. The kinetics of the thermosetting reactions may be affected by the differences, changing the way the resin responds to heat and pressure during processing. The control system adapts the cure schedule to account for these differences. If the constituent differences speed up or slow down the reaction, the system still controls the cure based on the physical state of the resin. Thus, accountability to variation of the material is maintained.

If, however, the composite part's shape was modified, in a design change for example, it would affect the control algorithm. Resin flow properties and heat transfer properties depend on the geometry of the part [12]. Changing the part's shape would alter the flow properties but certainly not the transducer signal, and one could not expect the same response of the resin from the applied cure parameters. Depending on how drastically the shape was changed, the algorithm would have to be altered or redeveloped, requiring more effort. The algorithm is part-specific and not universally applicable. In the future, the theory and analysis of the curing process may be in place so that an empirical approach is not necessary.

Temperature Control

Optimizing the cure cycle by using aggressive heating rates and eliminating intermediate dwell periods decreases the length of the cycle. But heating the laminate too fast can have a harmful effect on its quality. As the heating rate increases, the gel point of the resin is reached quicker, and, if consolidation of the fibers by the resin is not complete by the time the gel occurs, resin-rich zones are left inside the part. This problem can and should be addressed in conjunction with the pressure application strategy of the algorithm. The pressure affects resin flow also, but the use of the example here is to illustrate the motivation for controlling the temperature.

AFWAL has developed an APC system for epoxy resins that actively controls the temperature. The epoxy crosslinking reactions are exothermic so they give off heat, and the heat provides more energy for subsequent reactions, increasing the rate of the reaction. If the resin is given enough heat energy, it may reach an uncontrollable reaction rate and proceed to a premature gel. This phenomenon is called an *exotherm*.

The strategy of the AFWAL project is to control the temperature in a way that heats the laminate aggressively yet prevents an exotherm. The second time derivative of the temperatures in the top and center of the laminate is monitored.⁵ Very generally, using this information one can expose the point in time when the crosslinking reactions of the center are proceeding too rapidly, and cool the laminate to prevent the exotherm. The benefits of this temperature control strategy are fast-heating rates with protection from resin-rich zones caused by gel before adequate flow and the use of inexpensive thermocouples for control information. Still, the goal is to control the flow of the resin. Because the flow characteristics of the resin are described by the viscosity, a direct measurement of the viscosity would be more desirable, but also more difficult.

Another temperature control strategy for epoxy resins, proposed by Kranbuehl [3], involved monitoring the viscosity at the center ply and the surface ply of a flat composite laminate through dielectric measurement. The laminate was heated rapidly, and initially the viscosity at the top surface was less than at the center. This is because heat must be conducted from the surface to the center. The viscosity at the center later catches up to, or gets as low as, the viscosity at the surface by using extra heat produced in the exothermic reaction. At this point, the autoclave temperature is reduced to prolong this equal viscosity. Once the viscosity in both the center and surface goes through the second viscosity minimum, the temperature is raised rapidly to the final cure temperature and held there until completion.

In a temperature control strategy for a polyimide resin, General Dynamics/Pomona Division is "artificially" controlling the temperature in their APC system [2]. The optimum temperature profile was set in the laboratory by a trial and evaluation process. It was very aggressive in that it used no intermediate dwells at lower temperatures, only a ramp-and-hold strategy. The same profile will be used on the production floor, but the other events (pressure application and end of cure) will be manipulated from dielectric information. There is obvious simplicity in this "hard" temperature profile and it is part of their APC system strategy, but it is not controlling temperature as much as it is setting the optimum temperature and controlling the other cure parameters.

Pressure Control

It has already been mentioned that the applied pressure of processing equipment affects the flow of the resin. If the pressure is applied too soon or too late, there may be resin-starved or resin-rich zones as a consequence. In either case, a low-quality laminate is produced. The timing of the application of pressure has been heuristically developed (trial and error) in most of the research on APC.

In the APC system developed at AFWAL, the pressure is applied as soon as the dielectric transducer senses that flow has begun. It must be remembered that this strategy is for an epoxy resin. When processing with polyimide resins, some volatiles are always evolved after the pressure application. This makes the pressure application a more sensitive operation than when using an epoxy. In the General Dynamics APC algorithm, the pressure is applied at the viscosity minimum. The viscosity minimum is determined by the ionic conductivity signal's

⁵ F. Abrams, Air Force Wright Aeronautical Laboratories, Wright-Patterson Air Force Base, OH, personal communication, June 1989.
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maximum This method was found to work in the laboratory and verified in the production equipment Again, the technique has been developed empirically

By accurately predicting the end of cure, the expense of cure cycles that are too long can be avoided The overcuring or undercuring that produces a poor quality laminate can be avoided as well Near the end of cure the thermosetting chemical reactions are practically complete, and the molecular arrangement is no longer drastically changing From the physical basis of the sensors, an unchanging material corresponds to an unchanging signal, if all other parameters are also constant Because this event represents a slope and not a magnitude, it is an easy signal to distinguish This helps to make the end of cure the easiest cure event to predict

Examples of predicting end of cure from a microdielectric signal exist in the literature [2-4] They are all fundamentally the same in that when the ionic conductivity (σ) is no longer changing in time, the cross-linking reactions are complete and the cure is over Other conditions are required in the control law to ensure the actual physical state of the resin has been achieved and the signal is not the result of some malfunction For instance, part of the algorithm developed at AFWAL requires, in addition to a constant conductivity, that the temperature in the coolest spot of the laminate must be above a certain minimum temperature before ending cure⁵

Researchers have sought to prove or disprove a direct relationship between conductivity and degree of cure Many believe that dielectric techniques can predict the time at which cure is complete This accepted prediction should be exploited now, and as the understanding of dielectric measurements becomes more complete, it will no doubt add to the sophistication of process control methods

Sensor or other equipment failure can lead to the ruin of an expensive composite part if the APC system has nothing to fall back on in the absence of physical information from the transducer The obvious safeguard to this situation is to have a backup system if the first fails (fail-safe) No data exist on the reliability of any of the transducers currently on the market, but equipment failure is a problem being addressed by the people trying to develop an APC system for production⁶

How much to back up a system is largely a question of economics, each microdielectric sensor costs roughly \$30, but the cost of a scrapped part can quickly climb well beyond this In the APC system being developed at General Dynamics/Pomona Division, two identical parts are cured in the same machine, each one having its own transducer The signals are multiplexed and, in the event of one's failure, the system relies on the information of the other to complete the cure cycle

A zone exists in the cure schedule within which application of a certain cure parameter must be made to have any possibility of producing a satisfactory part This zone, called an *event window* [2], represents a time-temperature space below which an event cannot be triggered, and above which it must be triggered The algorithm initiates the event only within the window, because this window represents the first and last reasonable chances of getting a good-quality laminate

The windows are defined empirically in the laboratory If the monitoring equipment does not function properly, or quits altogether, the controlling algorithm relies on the window to trigger cure events in the absence of any sensor information These time windows should not be thought of as optimum moments to trigger events, the optimum event time is still triggered by information from transducers within the curing laminate Once the cure starts, there is no turning back, and an alternative strategy is more appealing than scrapping the part if the APC fails

Conclusions

The microdielectric-based transducers for APC have the most mature technology. There is no direct relationship between microdielectric measurements and the viscosity or degree of cure, but there is an interdependence that gives beneficial information from which to control the cure. More research needs to be done on the other two sensors capable of APC. More insight on the signals of the fluorescence optrode and acoustic wave guide is needed to be able to tell cure events from non-cure events and to be able to make signals reproducible.

Both these sensors offer a less expensive alternative to microdielectric-based sensors. The fluorescence optrode would be the easiest to implement in the production setting because the transducer is embedded in the tool and would not interfere with layup of the laminate. No sensor involves monitoring a molecular quantity, such as depletion of a functional group or change in molecular weight. Work continues in the development of all three.

An APC system, using microdielectric transducers, is ready for implementation in a production environment at General Dynamics/Pomona Division for processing polyimide laminates. Another system close to implementation is the Air Force Wright Aeronautical Laboratory's Qualitative Process Automation (QPA) project, which also uses microdielectric techniques.

The algorithms control any combination of these three major events: temperature, pressure, and end of cure. There are a number of different strategies, all empirically developed. They have been shown to work, producing good quality laminates with 50 to 70% reductions in cure times. Although they work, they are specific to part and resin, and there is considerable cost to refine and verify the system. Closed-loop cure control has not reached the stage where small manufacturers can produce resin-based advanced composites using black box APC systems with little associated development costs. Companies such as Micromet and Advanced Polymer Technology are working on turn-key systems with low learning curves, which would help to make this come true, but the state of the art is not there yet. Eventually, advancements in theory and analysis will replace the empirical basis for algorithm development.

Acknowledgments

This work was sponsored by Dow Chemical, USA. We would also like to thank the authors who provided additional resources for this work with their personal conversations.

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Analysis of Composite Material Containing Defects

REFERENCE. Frankle, R. S., Jones, D. R., Roberts, B. L., and Shusto, L. M., "Analysis of Composite Material Containing Defects," *Composite Materials: Testing and Design (Tenth Volume)*, ASTM STP 1120, Glenn C. Grimes, Ed., American Society for Testing and Materials, Philadelphia, 1992, pp. 320–329.

ABSTRACT: This paper describes work performed by Failure Analysis Associates, Inc. (FaAA) to develop capabilities for quantitative analysis and evaluation of parts containing defects. The work was conducted for Boeing Aerospace under the Astronautics Laboratory NDE Data Application Project (F04611-86-C-0080). In this project, FaAA used a finite element structural analysis code to predict the response of carbon-carbon involute rings that were intentionally fabricated with wrinkles and overstaged plies. Analytical techniques for modeling the effects of the defects were evaluated. The finite element analysis predictions compared favorably with data from hoop tension tests of the rings. A computer code was developed to use digital nondestructive evaluation (NDE) data automatically to define spatially varying material properties required by finite element analysis. The code provides a practical mechanism for automatically including the effects of defects in the analysis of actual hardware. The results of this effort demonstrate promising capabilities for the analysis of composite parts containing defects.

KEYWORDS: nondestructive evaluation, finite element analysis, composite materials, defects, wrinkle, dry ply

Historically, the evaluation of parts containing defects has largely been qualitative. Acceptance criteria are typically based on numbers and distribution of various types of defects. With the maturing of digital nondestructive inspection techniques and the improvement of analytical methods, it is now possible to postulate a more quantitative, part-specific assessment of defects in hardware. In this approach, the part evaluation includes an analysis that accounts for the unique, nondestructively measured characteristics of a particular part. Achievement of this capability will require improvement in defect detection methods, development of accurate analytical models for defects, creation of needed software, and integration of the capabilities into a coordinated, practical system. Work performed by FaAA in this project represents initial developments toward meeting this objective. This paper describes efforts to evaluate defect models, to use NDE data in finite element analysis, and to demonstrate the capability to analyze parts containing defects.

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Analysis of Defected Rings

To evaluate the effect of common defects on composite structures, FaAA analyzed the response of carbon-carbon involute rings containing intentional defects. The accuracy of the defect models was assessed by comparing the analytical predictions with the results of laboratory testing.

Material Description

Carbon-carbon involute composite cylinders were fabricated by Kaiser Aerotech for use in this project. The cylinder dimensions were nominally 9 in. (22.9 cm) long by 7.4-in. (18.8-cm) inner diameter by 0.5-in. (1.3-cm) wall thickness. The involute architecture was designed with a helix angle of 0° and an arc angle of 5.6° at the inner diameter and 4.9° at the outer diameter. One cylinder was fabricated without intentional defects to represent nominal material. During the fabrication process, defects were introduced intentionally into two cylinders to simulate types of defects observed in actual parts, namely dry plies and wrinkles. The fabrication process for each defect type is summarized below.

Dry Ply Cylinder—In the past, radiographically detected low-density regions have been attributed to so-called dry or resin-starved plies. To simulate this condition, two groups of contiguous involute plies were overstaged before incorporation in one cylinder, referred to as the dry ply cylinder. One group consisted of three overstaged plies, and the other contained five overstaged plies. In both cases, the nominal ply thickness was 0.013 in. (0.033 cm). The two groups of overstaged plies were inserted 180° apart in the involute cylinder.

Wrinkle Cylinder—A triangular bar was pressed into the outer diameter (OD) of the cylindrical preform at two circumferential locations. The bar extended along the cylinder's entire length. The result, in the finished wrinkled cylinder, were two wrinkles in the hoop-radial plane 180° apart. One wrinkle was approximately 0.1 in. (0.25 cm) high in the radial direction by 0.8 in. (2.0 cm) long in the circumferential direction. The larger wrinkle was approximately 0.2 in. (0.5 cm) high radially by 0.8 in. (2.0 cm) long circumferentially. Following densification, the indentations on the outer diameter of the wrinkled cylinder were machined off, reducing the wall thickness of this one cylinder to 0.35 in. (0.89 cm) and the involute arc angle at the OD to 5.1° . A photomicrograph of the large wrinkle is shown in Fig. 1.

Finite Element Analysis

FaAA performed finite element analysis of rings removed from the cylinders and subsequently used as laboratory test specimens. The nominal dimensions of the rings are listed in Table 1.

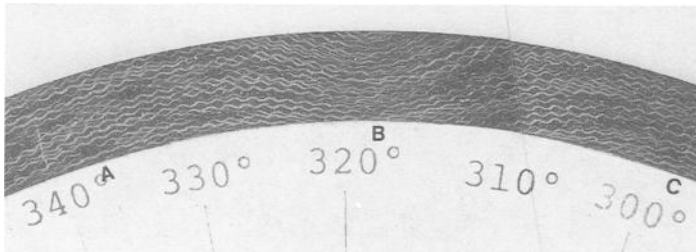


FIG. 1—Photomicrograph of large wrinkle.

TABLE 1—Ring dimensions.

Ring	ID, in. (cm)	OD, in. (cm)	Length, in. (cm)
Nominal	7.4 (18.8)	8.4 (21.3)	0.5 (1.3)
Dry ply	7.4 (18.8)	8.4 (21.3)	0.5 (1.3)
Wrinkle	7.4 (18.8)	8.1 (21.3)	0.5 (1.3)

Finite element models constructed of the rings contained 600 4-node isoparametric quadrilateral elements equally spaced as 120 in the circumferential direction and 5 in the radial direction. The finite element mesh was designed to correspond to the cylindrical geometry of the rings without trying also to align the mesh to the involute architecture of the material.

A plane strain analysis of room temperature internal pressure loading to 100 psi (689 kPa) was conducted using the FANTASTIC® code. FANTASTIC is an acronym for Failure Analysis Nonlinear Thermal and Structural Integrated Code. Because the tests were conducted at room temperature, only the finite element structural analysis module, FAST, was used. Linear elastic material properties were used in which the tensile and compressive properties were assumed equal.

Defect Models

The finite element models were modified in the region of the defects to account for the expected material changes at these locations. For the wrinkle ring, the model accounted for the effect of the wrinkles on local ply orientation. For the dry ply ring, the model incorporated a local reduction in material properties due to the postulated lower density in the dry plies. A description of each finite element model follows.

Wrinkle Ring—For the wrinkle ring, the two wrinkles in the hoop-radial plane were located 180° apart, as shown in the finite element model (Fig. 2). To account for the presence of the wrinkles, the material properties in the finite elements in the wrinkled regions were trans-

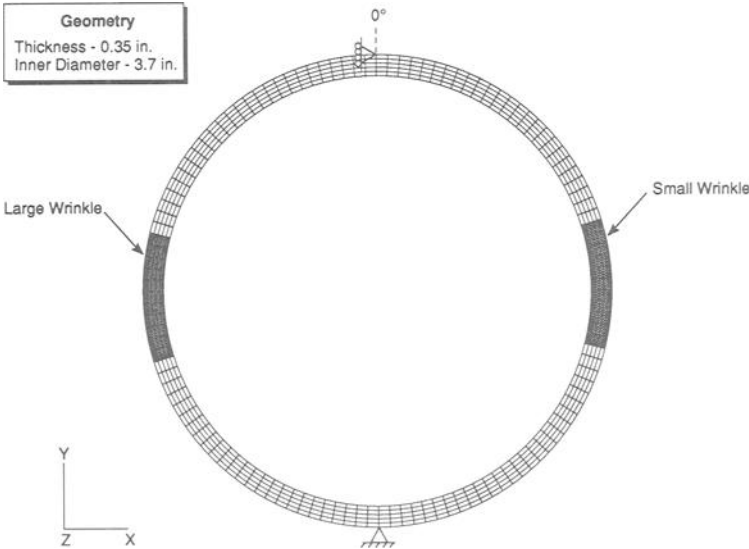


FIG. 2—Finite element model of wrinkle ring.

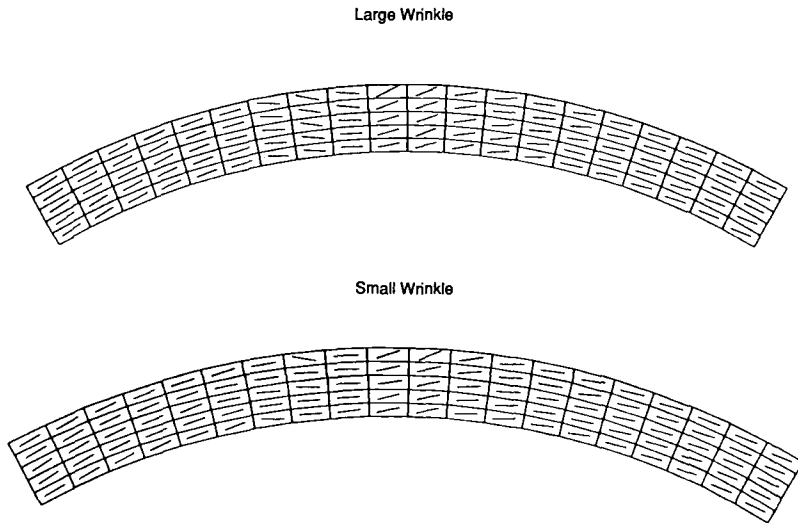


FIG. 3—Wrinkle ply orientations.

formed to correspond to the local ply orientation. Classical composite transformations were used to transform the constitutive properties (elastic moduli) and failure properties (strengths). The local ply orientation in the wrinkles was obtained by overlaying the finite element mesh onto photomicrographs of the actual wrinkles and manually determining the nominal ply orientation in each finite element. The resulting orientations for both the large and the small wrinkle are shown schematically in Fig. 3. These orientations were input to the transformation equations, which modified the local nominal material properties accordingly. The transformation of material properties to the local ply orientation was the only modification made from the analysis of the nominal ring.

Dry Ply Ring—For the ring with the overstaged plies, the presence of the dry plies was accounted for by reducing the material properties in the finite elements in the region of the dry plies. As shown in Fig. 4, two different linear variations with density were employed, depend-

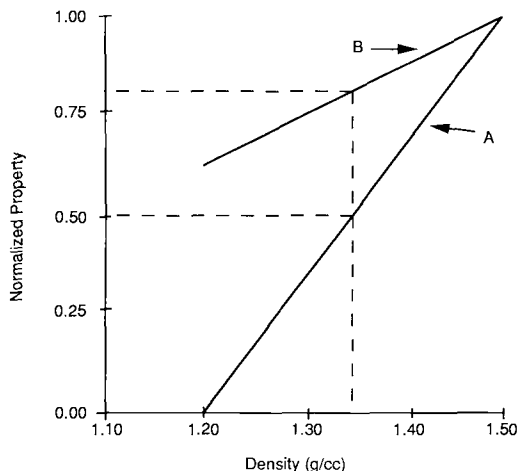


FIG. 4—Dry ply material property reduction.

ing on the specific material property. The following matrix-dependent properties were assumed to decrease to zero at a density value low enough that the matrix phase is ineffective:

- All elastic stiffnesses.
- Strengths in interlaminar shear and cross-ply tension.
- Compressive strengths in warp and fill directions.
- Warp-fill shear strength.

This variation with density corresponds to Line A in Fig. 4. The following fiber-dependent properties were assumed to decrease to corresponding fiber properties in the absence of an effective matrix phase:

- Tensile strengths in warp and fill directions.
- Compressive strength in cross-ply direction.

This variation with density is shown as Line B in Fig. 4. Poisson's ratio was assumed to be constant with changes in density.

Actual low-density regions observed in X-ray films exhibit a 10% reduction in density from the nominal value for this material. To reflect this density reduction, the material properties in the region of the overstaged plies were reduced to either 50 (Fig. 4, Line A) or 80% (Fig. 4, Line B) of the nominal values, depending on the specific property and its assumed relationship to material density.

The finite element model of the dry ply ring is shown in Fig. 5. The model included only the group of five overstaged plies, which was considered the worst case dry ply defect. Note that the elements representing the overstaged plies, in which the material properties were reduced, approximate the involute ply pattern. The dry ply region was modeled as extending from the OD through 60% of the ring thickness, as shown in Fig. 5.

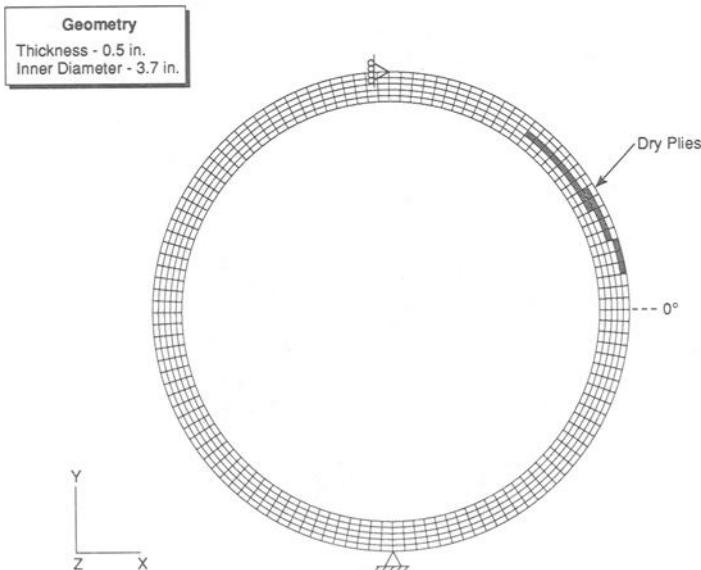


FIG. 5. Finite element model of dry ply ring.

Analytical Predictions

The effect of the wrinkles on the predicted factor of safety is illustrated in Fig. 6. Figure 6 presents the circumferential variation in the three-dimensional (3D) Tsai-Wu factor of safety at five different radial locations in the wrinkle ring. In the Tsai-Wu formulation,

$$F_{ij}\sigma_i\sigma_j + F_i\sigma_i = 1 \text{ for } i, j = 1, \dots, 6$$

the interaction terms were defined as:

$$F_{ij} = -\frac{1}{2}(F_{ii}F_{jj})^{1/2}$$

where

$$i \neq j \text{ and} \\ F_{ij} = 0 \text{ for } i, j = 4, 5, 6$$

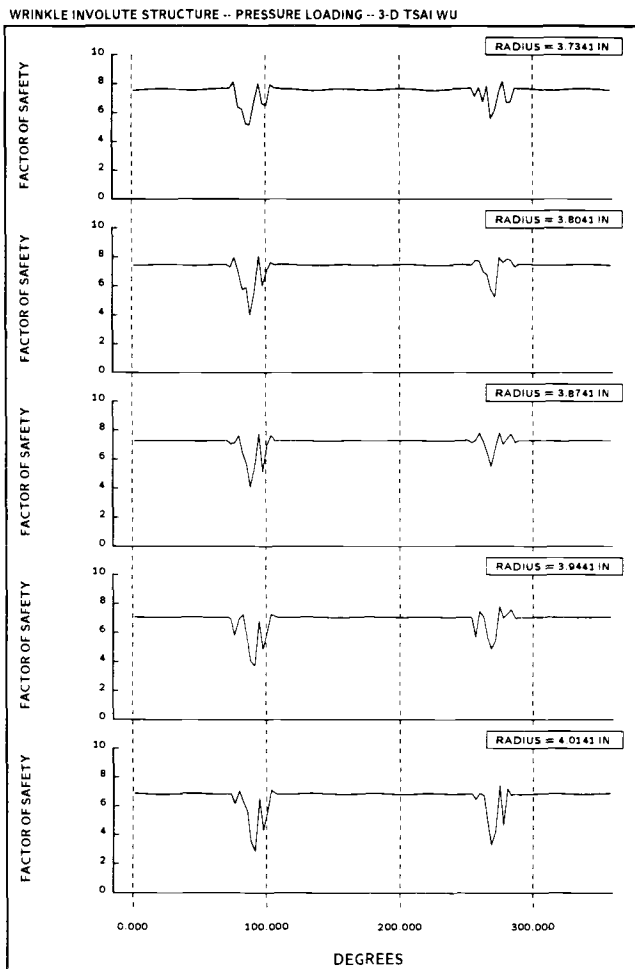


FIG. 6. Wrinkle ring—predicted factor of safety.

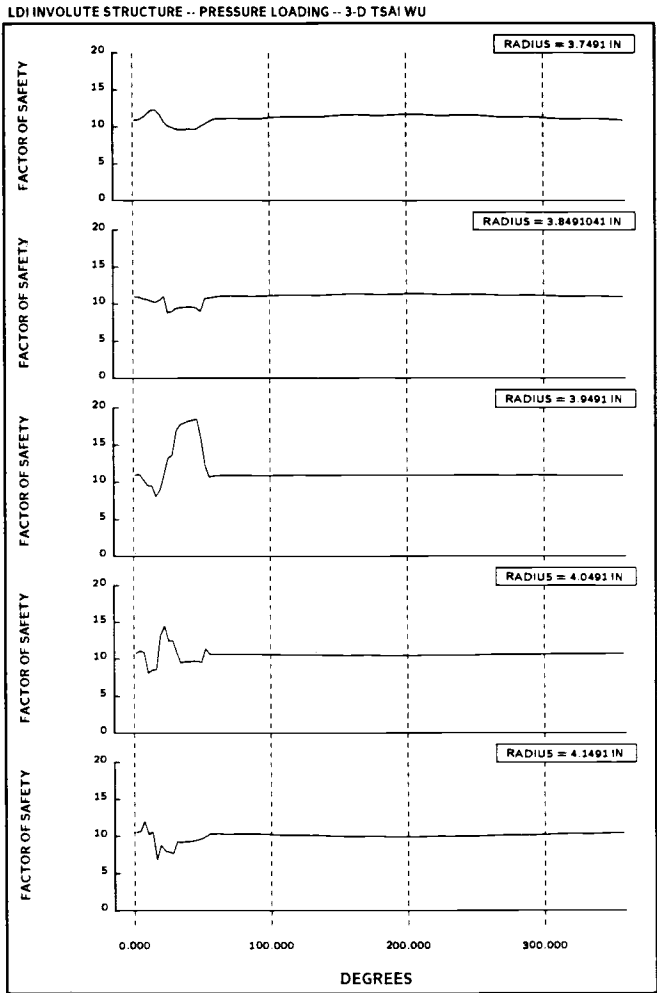


FIG. 7—Dry ply ring—predicted factor of safety.

This assumption was made in the absence of biaxial data for this material [1]. As shown in Fig. 6, there is a significant reduction in the factor of safety as a result of the presence of the large wrinkle at 90° and the small wrinkle at 270°. A similar result is shown for the dry ply ring in Fig. 7.

Comparison with Test Data

The nominal, wrinkle, and dry ply rings analyzed by FaAA represented hoop tension test specimens removed from the involute cylinders. Southern Research Institute tested two of each type of ring to failure in hoop tension. Because the analyses were linear, the analytical predictions for 100-psi (689-kPa) internal pressure could be scaled according to the predicted margin of safety to obtain the predicted failure pressure. The predicted pressure was then compared with the reported test results (Table 2). Table 2 lists the ratio of predicted failure pressure to average measured failure pressure.

TABLE 2—*Ring results*

Ring Type	Normalized Failure Pressure ^a
Nominal	1.17
Dry ply	1.02
Wrinkle	1.03

^a FANTASTIC prediction divided by average of two test measurements

Discussion

Although this effort was limited to two test specimens of each type, results illustrate that the approaches taken for including the effects of dry plies and wrinkles in finite element analysis may be valid first-order approximations. The transformation of properties to reflect the wrinkle ply orientation appears to be a fundamentally reasonable approach. The accurate agreement with test data obtained in this exercise tends to support this approach, originally postulated by Jortner [2]. Note that the least accurate prediction was for the nominal rings. This may be because the assumed failure criterion was less accurate without the presence of a known defect. In fact, the tested nominal rings may have failed prematurely at an unintentional defect that was not detected and not accounted for in the finite element analysis. This would explain the analytical overprediction of failure pressure for the nominal rings.

Use of NDE Data in Finite Element Analysis

One approach to assess quantitatively the effect of defects on the performance of a specific part is to analyze the particular part, including in the analysis the effects of any anomalies that may be present. This requires that the information on defects and other part characteristics derived from nondestructive evaluation (NDE) be input to the finite element analysis. Under the NDE Data Application project, FaAA developed FANDEP (FANTASTIC NDE Preprocessor) software that automatically assigns NDE data to finite element integration points at the same spatial location. The NDE parameters can be used to modify the analytical model to account for the presence of nondestructively detected features and anomalies. The approach employed in the project was to modify the material properties in accordance with the measured NDE parameters.

FANDEP Overview

Figure 8 illustrates the approach for using NDE data in analysis. Figure 8 shows the two primary inputs to FANDEP:

1. A database of measured NDE parameters derived from digital NDE inspection, such as X-ray computed tomography (CT) and ultrasonic (UT) data.
2. An existing finite element model, which may have been previously created for design analysis.

For each integration point in the finite element model, FANDEP locates the NDE pixel at the same physical location and assigns the NDE parameters associated with that pixel to the integration point. A set of material properties required for the analysis is then assigned to the integration point, based on correlations between the properties and the NDE parameters. These properties are provided automatically as input to the FANTASTIC code. With this approach,

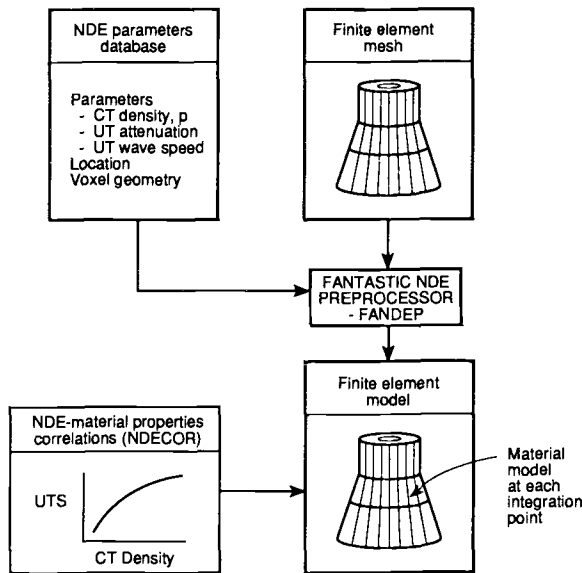


FIG. 8—Role of FANDEP preprocessor.

spatial variations in the part detected by NDE can be reflected in corresponding spatial variations in the material properties used in the analysis.

FANDEP currently accepts input from CT or UT data or both. The particular equipment used to generate the data is not important, only the format of the data as input to FANDEP. To account for 3D variations in a part, FANDEP operates on a fully 3-D finite element model. It currently supports two elements in the FANTASTIC code, either an 8-noded or 20-noded brick element. FANDEP has built-in capabilities to detect and account for missing data and duplicate data. FANDEP also deals with differences in geometry between a specific part and the finite element model, which is based on the engineering drawing. Because of geometry differences between the part and the model, integration points in the model, especially those near the edges, can lie outside the boundary of the part. Therefore, FANDEP includes an algorithm that maps the finite element model into the actual part, where the part geometry is defined by the CT data. All of the integration points are still located correctly within the elements, but the elements are now positioned properly within the boundaries of the part. This operation is performed automatically before assignment of NDE parameters at each integration point.

FANDEP is designed to operate in an engineering workstation environment compatible with the current trends in both NDE and engineering analysis. The code is modular in design and written in the C computer language for the Unix operating system. The portable and expandable modular design accommodates operation on different machines or future code development.

In this effort, FaAA used FANDEP to input NDE data on the composite cylinders into finite element analysis. The ASCII data consisted of 88 megabytes of CT density data and 8 megabytes of UT attenuation data. The 3D finite element model of the cylinders contained 3600 8-noded brick elements. Once the NDE data were put into FANDEP input format, the FANDEP operations required 10 min on a SUN 3/180 workstation. Proper function of the code was verified by comparing the measured distribution of CT and UT data with the distribution in the finite element model resulting from FANDEP. The success of this experience provides

an example of the practical capability of FANDEP to handle the typically large data sets resulting from current digital NDE inspections

Conclusions

The results of this effort demonstrate promising capabilities for the analysis of parts containing defects. Analytical models for wrinkles and dry plies used in the analysis of intentionally defected composite involute rings provided results that compared favorably with test measurements. Additional analysis and testing are required to evaluate further the analytical approach and the range of its validity.

The analytical approach that uses NDE data in finite element analysis appears to provide a viable basis for an automated procedure to evaluate parts containing defects. This approach of modifying properties based on NDE parameters currently lends itself quite well to defects that can be detected with existing NDE techniques and modeled by simply changing material properties. For these defects, there is a need to pursue correlations between NDE parameters and material properties used in analysis. However, in the case of a wrinkle, there is no proven technique for nondestructively measuring ply or reinforcement orientation. Therefore, in spite of the fact that the analytical results tend to validate the modeling approach, use of this approach for wrinkles is hampered by the lack of a reliable NDE detection and measurement technique. Work at FaAA in eddy current inspection of carbon-carbon materials is showing promise for practical nondestructive measurement of reinforcement direction.

The FANDEP code is designed to assign NDE parameters accurately to finite element integration points. The code also operates in an engineering workstation environment. Further work is required to exercise the code and evaluate its application to a wider variety of NDE data sets, parts, and computer systems. Further expansion of FANDEP's capabilities will include application to non-axisymmetric parts, inclusion of other digital NDE techniques, and faster run times.

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Image Enhancement Techniques for Ultrasonic NDE Applications

REFERENCE: Steiner, K. V., "Image Enhancement Techniques for Ultrasonic NDE Applications," *Composite Materials: Testing and Design (Tenth Volume)*, ASTM STP 1120, Glenn C. Grimes, Ed., American Society for Testing and Materials, Philadelphia, 1992, pp. 330–343

ABSTRACT: Ultrasonic nondestructive evaluation (NDE) methods are reliable in the inspection and certification of composite structures. Complex structures are scanned by sophisticated automated multi-axis robotic systems. The acquired data are processed to generate images based on peak-amplitude or time-of-flight data. Digitizing of the ultrasonic information has opened the door to powerful computer graphics image enhancement techniques.

Image enhancement techniques can be applied such that critical information is highlighted and defects detected in the structure are classified. Used properly, image enhancement of digitized data dramatically improves the interpretability of ultrasonic information. A wide variety of image enhancement techniques, applicable in particular to ultrasonic NDE, are discussed. In addition, a new method of interpreting ultrasonic data by working from digitized full-volume waveform information is presented. This work, combined with statistical analysis methods, enables the operator to perform complete post-scan evaluation.

KEYWORDS: composites, nondestructive evaluation, ultrasonics, image enhancement

Two related methods for ultrasonic nondestructive evaluation (NDE) are in use today: the through-transmission method and the pulse-echo method (Fig. 1). With the through-transmission method, an ultrasonic pulse is emitted from a transducer and travels through water, which acts as a coupling medium, before it penetrates the composite specimen. The ultrasonic wave is attenuated while proceeding through the specimen, and the remaining transmitted energy is received by a second transducer positioned on the opposite side of the sample. The pulse-echo method, in contrast, incorporates only a single transducer, which emits the ultrasonic pulse and receives the reflected signal energy that returns from the specimen [1]. In either case, it is crucial to position the transducers precisely to receive the transmitted or reflected energy, respectively.

A comparative study of through-transmission and pulse-echo examination of a graphite-epoxy panel is shown in Fig. 2 with the pulse-echo data in the right half of the figure. The scanned panel has varying thicknesses of 16, 32, 48, and 64 plies, increasing from the top to the bottom. Teflon® shims and other intentionally implanted flaws of various sizes and depths show up in either ultrasonic technique. In the through-transmission scans, delaminations show up as dark areas of lower-than-average intensity; in the pulse-echo scans, they show up as light areas of higher-than-average intensity. It is also obvious that the through-transmission method is not as dependent on the varying depth as the pulse-echo mode because in through-transmission mode the waveform travels only once through the material before reaching the receiver.

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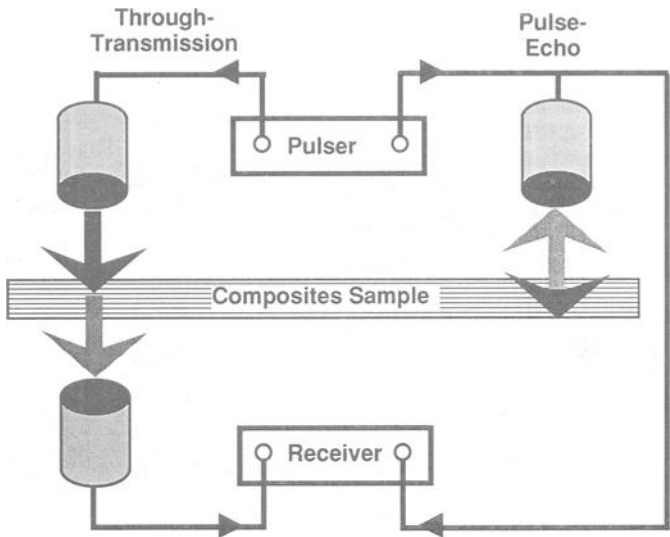


FIG. 1—Principles of ultrasonic through-transmission and pulse-echo methods.

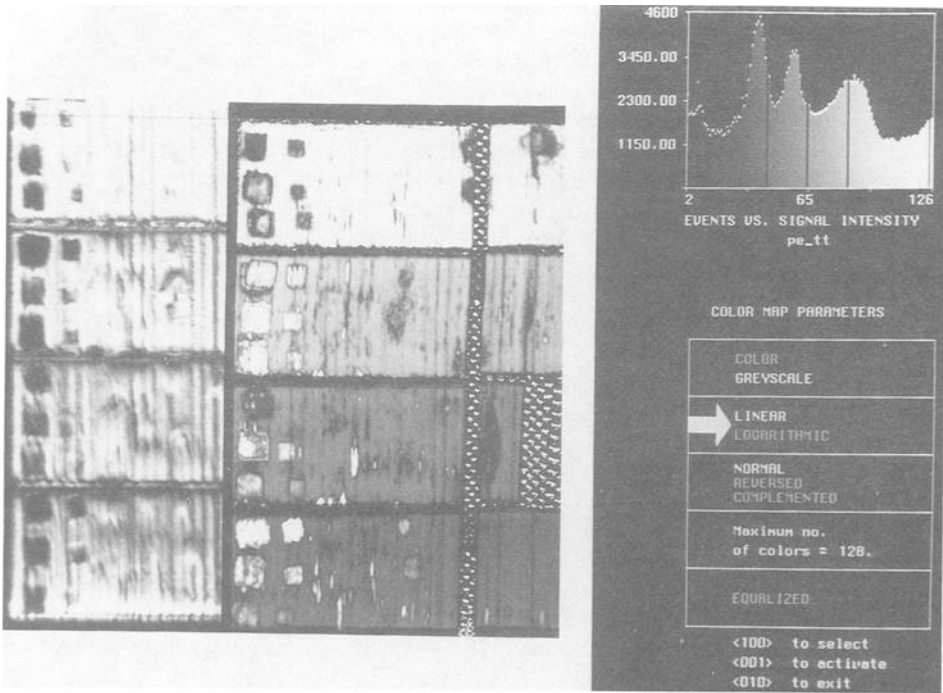


FIG. 2—Results of ultrasonic through-transmission (left) and pulse-echo C-scans.

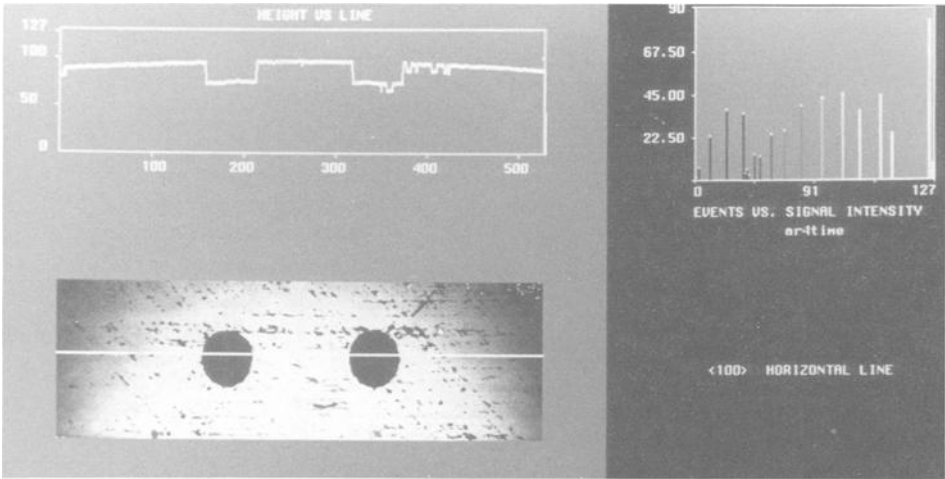


FIG. 3—Time-of-flight data of 64-ply graphite-epoxy sample with implemented Teflon shims.

The received waveforms are preprocessed in an ultrasonic analyzer where the maximum amplitude within selected time intervals, also known as gates, is computed for each pulse. These values are displayed as pixels in gray-scale or pseudo-color on a computer graphics screen, and the resulting two-dimensional picture is known as a C-scan. In addition, time-of-flight data can be computed for pulse-echo scans, which give an indication of the flaw depth or specimen thickness based on the time difference between the reception of the echo from the specimen's front surface and the flaw or back surface, respectively. Time-of-flight data and a corresponding depth-profile for a 64-ply graphite-epoxy specimen with two implemented circular Teflon shims simulating delaminations are shown in Fig. 3.

The flaws found in a composite specimen are then categorized in defect classes. Among the defect classes detectable with ultrasonic methods are voids, contaminations, and delaminations. Other features that can be categorized include matrix-rich and matrix-starved regions within the composite specimen. Changes in thickness and geometry may influence the signature of the ultrasonic echo, and the resulting images have to be compensated for these variations.

Ultrasonic NDE Facilities

Researchers at the Center for Composite Materials at the University of Delaware have developed a robotic ultrasonic inspection system for complex three-dimensional objects [2]. A six-axis PUMA 562 robot is suspended upside-down from a gantry above a 5- by 8- by 3-ft (1.5 by 2.4 by 1 m) immersion tank (Fig. 4). The scanning procedure is supervised by the central control unit, an Omnibyte 68000 microprocessor, which communicates with the robot controller while triggering the pulser of a Metrotek ultrasonic analyzer at preselected intervals. Ultrasonic waves are transmitted from the robot-mounted transducer and travel through the water into the composite part under inspection. The maximum amplitudes within up to three gated regions of the returned echoes are captured and digitized. The resulting C-scans are displayed on the screen of a Silicon Graphics IRIS 2400 Turbo, a Unix-based three-dimensional (3-D) graphics workstation with over 300 megabytes of hard-disk storage.

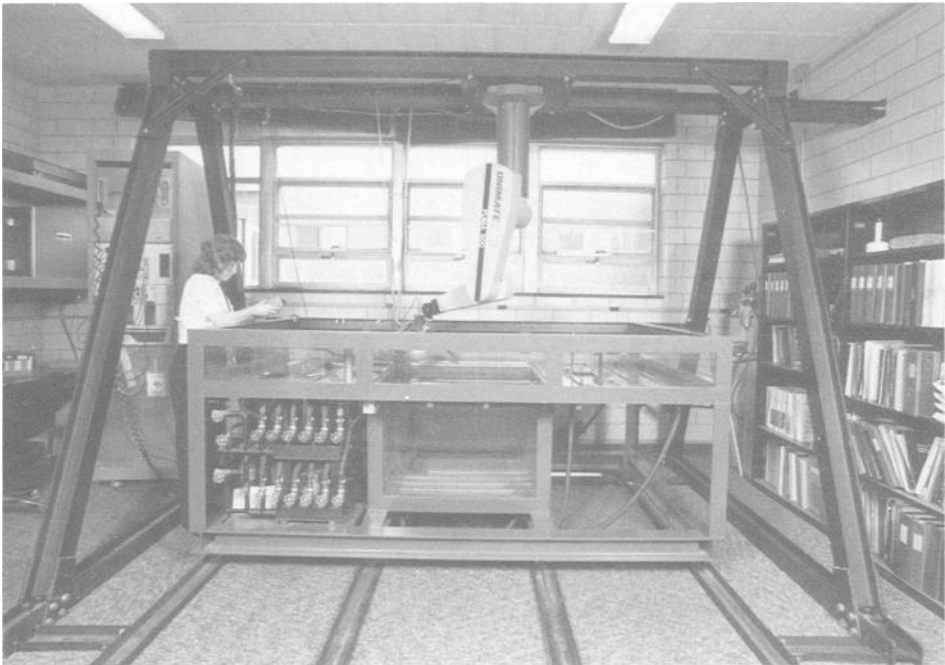


FIG. 4—Robotic ultrasonic nondestructive inspection workstation at the Center for Composite Materials at the University of Delaware.

The resolution of the scanning system is limited by two factors, the frequency of the transducer and the accuracy of the robotic manipulator. Within the Center, the most commonly used transducers have frequencies ranging from 15 to 25 MHz, resulting in a maximum resolution of between 0.075 and 0.05 mm. On the other hand, the robot's repeatability is limited to ± 0.1 mm. Therefore, the limits of either component are reached at about the same resolution of 0.1 mm.

In-house developed software to perform histogram-based image enhancement techniques is used to view and manipulate the C-scan results on the screen [3]. The histogram of the current image is displayed in the upper right corner of the screen, and menus with numerous options for powerful image manipulations are located below the histogram area. (Figures 3, 8, 9, 11, and 12 have been taken from this program.)

These facilities allow the inspection of parts with both types of digital C-scans, the pulse-echo and through-transmission method. In general, pulse-echo scans are preferred at the Center because they require only one transducer positioned normal to the surface of the part under inspection and because the reflected echo allows for determination of thickness and depth of a flaw based on time-of-flight data.

Image Enhancement Techniques

Powerful image enhancement techniques can aid the human observer in the detection of flaws in composite structures. The histogram-based image enhancement program developed at the Composites Center in Delaware features a wide variety of options [4]. Among the fea-

tures are achromatic and pseudo-color representations, contrast stretching, histogram equalization and color table forcing functions, color level slicing, Kernel enhancement, and fast Fourier transformation (FFT) analysis and filtering capabilities.

The intent of contrast stretching is to expand the narrow range of brightness values typically present in an image over a wider range of gray values. The result is a new image designed to accentuate the contrast between features of interest in the image. Figure 5 (top) shows an average unenhanced histogram [5]. Figure 5 (center) shows the same histogram after eliminating all values below 4 and above 64 and then contrast stretching, thus increasing the range of intensities available. Another way of increasing the image contrast is to equalize the histogram. In this case, the number of occurrences for each level is being distributed, resulting in a more equalized histogram. Figure 5 (bottom) shows the results of this operation performed for the histogram in Fig. 5 (top). The C-scan of a $[0/90/\pm 45]$ graphite-epoxy specimen shown in Fig. 6 has been subjected first to contrast stretching and then to equalization enhancement methods.

C-scan data are preferably displayed in gray-scale because the human eye is more susceptible to subtle differences in gray intensities than in color shades. Color is used to highlight specific features of the C-scan that are superimposed on the achromatic background. This method is also known as color level slicing, where certain sections of the histogram are highlighted in a user-selected color.

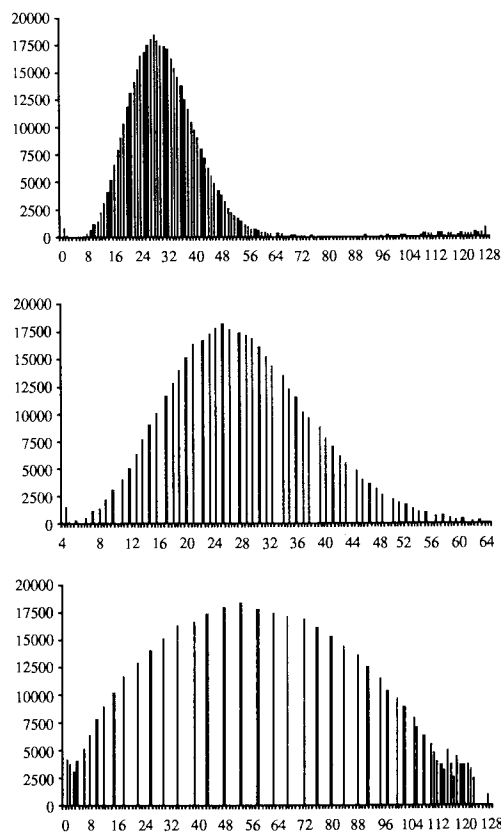


FIG. 5—(top) Original 128-level histogram without image enhancement, (center) histogram after contrast stretching, stretched between levels 4 and 64, and (bottom) 128-level histogram after equalization.

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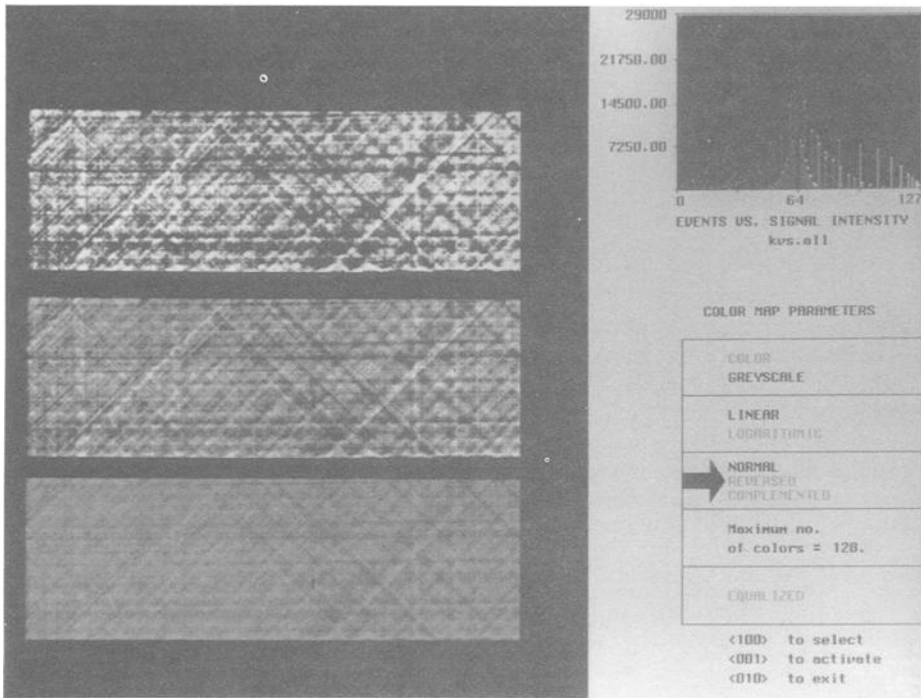


FIG. 6—(Bottom to top) Original C-scan of $[0/90/\pm 45]$ graphite-epoxy specimen, after contrast stretching, and after equalization.

Kernel enhancement, which includes such features as edge enhancement, focusing, smoothing, and noise reduction, is another powerful method for enhancing specific features in a C-scan. Specific Kernels can be applied, which lead to vertical, horizontal, or diagonal edge enhancements. In the latter case, the applied Kernel improves the visibility of the 45° fiber orientation by creating an optical illusion. This image-enhancing technique generates software “shadows” according to the intensity variation between two pixels, which helps the human inspector with the image interpretation by giving the illusion of a stereoscopic image. Applying a high-pass filter Kernel to the same C-scan data leads to improved contrast and to a sharpened image. Images with rough contrasts should be processed with a small Kernel, for example, 3 by 3 pixels, whereas images with smooth features should be processed with a large Kernel, for example, 9 by 9 pixels [5]. Some commonly used 3 by 3 matrix Kernels are shown in Fig. 7.

Figure 8 shows results of applying a variety of Kernels to an ultrasonic C-scan image. The original image is displayed on the bottom. The additional images have been Kernel-enhanced with diagonal difference in $+45^\circ$ direction, diagonal difference in -45° direction, and horizontal difference, as seen from bottom to top, respectively. The fiber orientation visibility in the Kernel direction is dramatically increased while other features in the image outside the Kernel direction are suppressed.

The most powerful technique available for image enhancement is fast Fourier transformation (FFT) [6]. Transforming the images into the frequency domain prepares the image information for filtering before invoking the inverse FFT. In addition, the frequency domain enables easy determination of the fiber orientation in the specimen (Fig. 9). This feature is especially useful for short-fiber composites, in which it is sometimes difficult to determine the fiber orientation and manual digitizing techniques are still in common use.

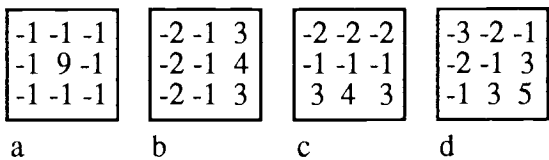


FIG 7—Common image enhancement Kernels for image processing: (a) high-pass filter, (b) vertical difference, (c) horizontal difference, and (d) 45° diagonal difference.

One of the difficulties in quality assurance is the transfer of information from the quality control department to the production or design department. It is critical that the information provided by the nondestructive evaluation group is easily understandable and can be interpreted by members of other departments. Image enhancement techniques such as the ones described above enable communication between these departments.

These image-processing techniques must be used by trained and qualified personnel because they cannot only enhance previously hidden defects but also, if not used properly, can suppress information about previously visible defects.

Statistical Histogram Analysis

A square panel, 305 by 305 mm, was laid up bidirectionally [0/90] with 32 plies of graphite-epoxy prepreg (IM6G/2220-3). The top section, a 190- by 305-mm large plate, was manufactured with a variety of defects embedded between the center plies before curing. These defects included backing paper, Teflon chips, acetone spills, and Frekote®, a mold release agent from

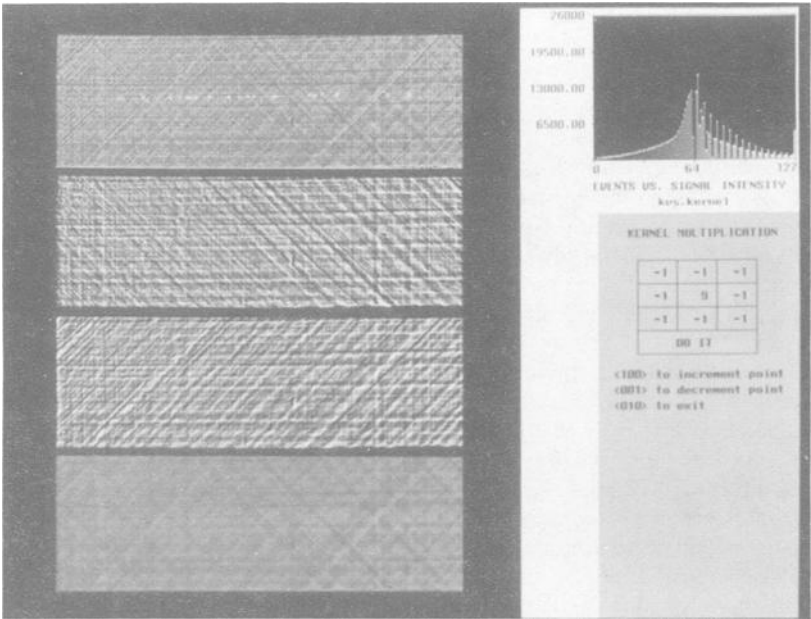


FIG. 8—(Bottom to top) Original C-scan of [0/90/±45] graphite-epoxy specimen, +45° diagonal Kernel, -45° diagonal Kernel, and horizontal Kernel.

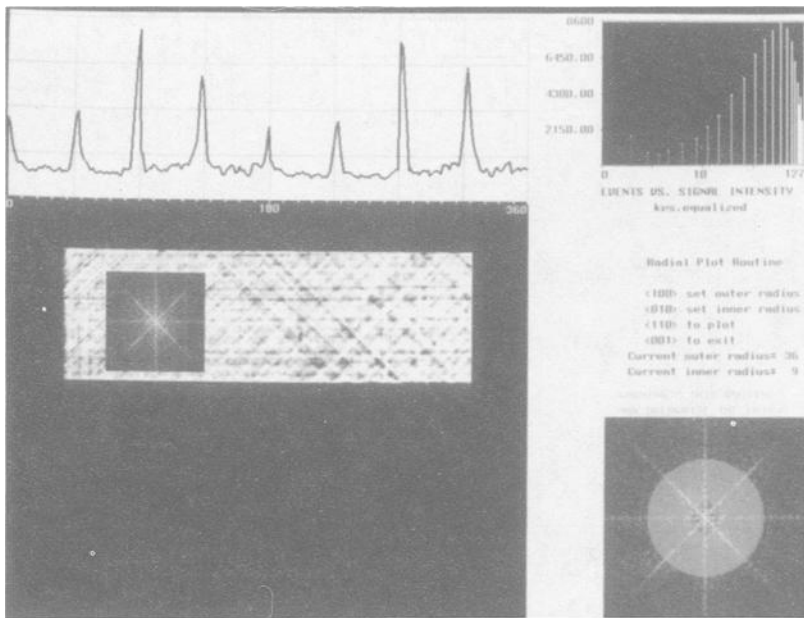


FIG. 9—Fast Fourier transformation of $[0/90/\pm 45]$ graphite-epoxy specimen with indication of fiber orientation.

Dexter Hysol, to simulate delaminations. Matrix plies and a punched hole in one of the prepreg layers were intended to enhance matrix richness, and sand particles were implanted to act as voids [7]. A schematic of the composite panel is shown in Fig. 10.

A 60-mm-wide section on the bottom of the sample was cut from the sample and used in a four-point static bending test where it was subjected to a bending stress of 2150 MPa, at which time a delamination occurs. A second, 55-mm-wide section was separated from the manufactured sample, and an 18-mm-diameter hole was drilled in its center to initiate the growth of a delamination during a multicycle four-point bending test. The test parameters were 200 000 cycles with a maximum bending stress of 2400 MPa at a frequency of 20 Hz.

Two C-scans were performed on this panel. First, the gating was selected around the back surface; the resulting C-scan image is shown in Fig. 11. The second C-scan was performed with the gating selected around the midsection of the waveform interface with the composite plate (Fig. 12). Each gate selection leads to the detection of some but not all implanted flaws. In addition, the flaws in the back-surface-gated scans show up as lower-than-average areas, whereas the flaws in the midsection-gated C-scan appear as higher-than-average intensities.

The shift in signal intensity is visible also in the image histograms, shown in Fig. 13 (top) for the back-surface gates and in Fig. 13 (bottom) for the midsection gates. The vertical lines mark the average signal intensity μ , with the middle line and the standard deviation measured from the average value $\mu \pm \delta$ as the two lines on either side of the average intensity for each histogram [8].

Statistical values of the image histograms for the back-surface-gated specimen are compared in Table 1. The various sections used for the analysis are shown in Fig. 10. These include an area without defects, an area with extra matrix plies, the area with implanted backing paper, and the area with the intentional Frekote spill. Preliminary results suggest that the most useful

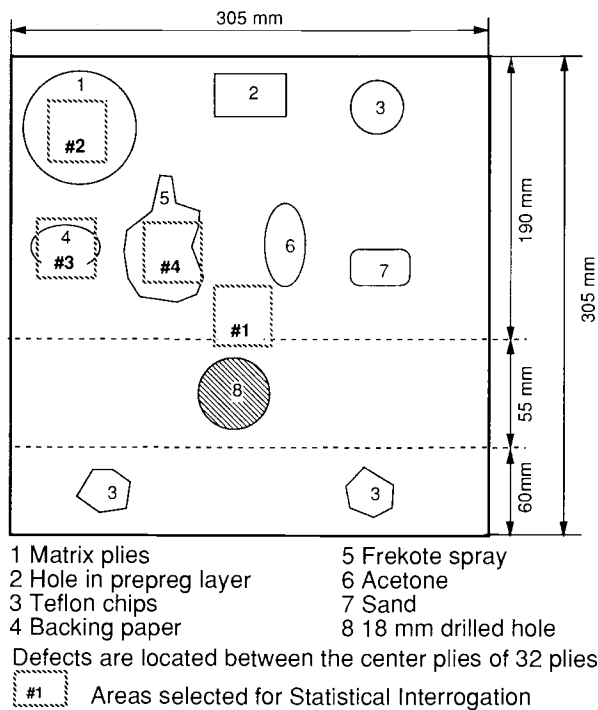
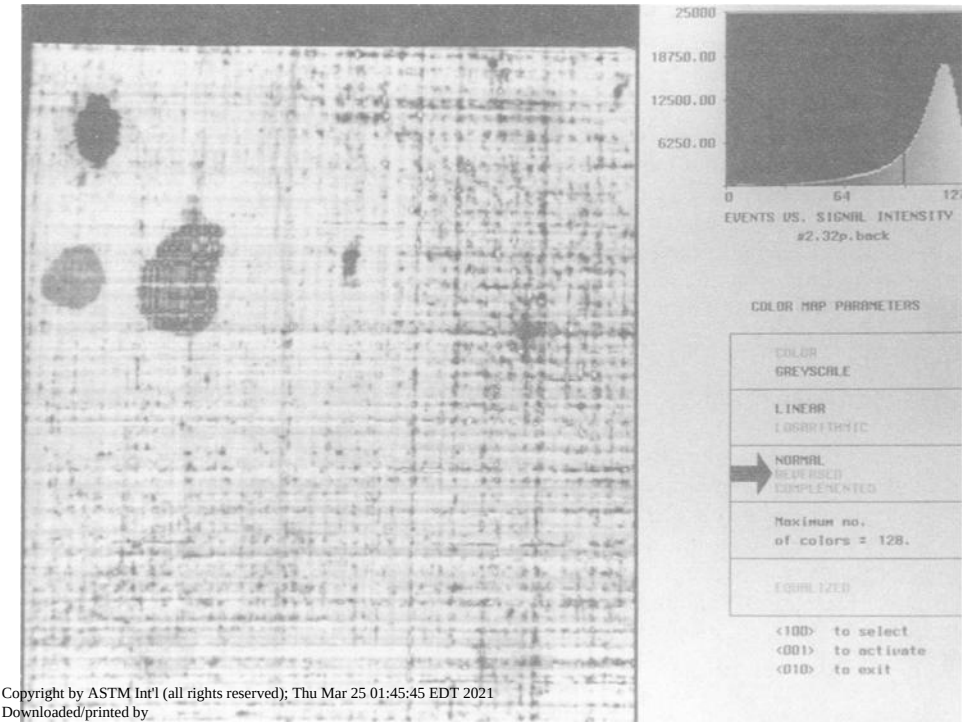


FIG. 10—Schematic of 32-ply graphite-epoxy specimen with intentionally implemented flaws.



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FIG. 11—Ultrasonic C-scan of 32-ply graphite-epoxy specimen gated around back surface. Size of specimen: 305 by 305 mm.

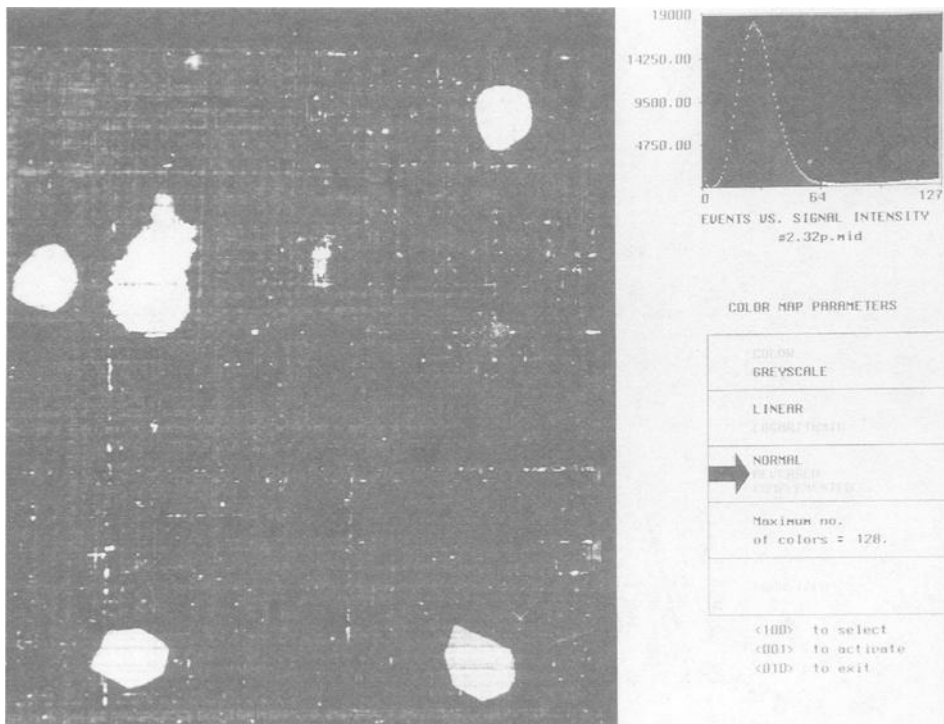


FIG. 12—Ultrasonic C-scan of 32-ply graphite-epoxy specimen gated around midsection. Size of specimen: 305 by 305 mm.

information can be obtained through the average level, the peak level, and the standard deviation. These values could be used to establish decision criteria about the overall performance of the composite part. Additional studies are needed to verify the results and to establish a scientific base for this type of statistical analysis.

Digitized Full-Volume Waveform Analysis

Digitized full-volume waveform scanning, a technique developed and implemented at the Center over the last four years, allows for digitization and storage of the complete waveform information at a rate of 100 to 500 bytes per ultrasonic pulse. The digitization is accomplished through a LeCroy 9400 digital oscilloscope at a sample rate of 100 MHz. With this scan method, no gates have to be chosen before the scan procedure, which opens the field to statistical waveform analysis and complete post-process image manipulation. However, implementation of this method requires a dramatic increase in number of recorded data per scan.

Another in-house developed software package is applied to interpret the results of the full-volume waveform scan [9]. An arrow-shaped cursor is used to select one of the sample slices shown in the bottom half of the screen, and the waveform corresponding to the current transducer position is displayed in the upper left quarter of the screen. Two vertical lines in the waveform display-field indicate the selected gate settings of the active sample slice. The corresponding histogram is generated and shown in the upper right quarter of the screen.

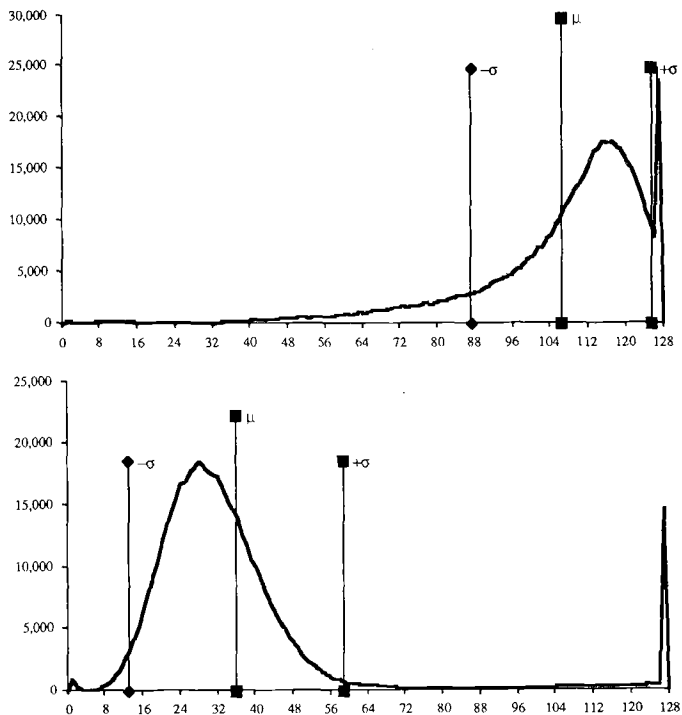


FIG. 13—(top) Histogram for 32-ply specimen C-scan with back-surface gate and (bottom) histogram for 32-ply specimen C-scan with midsection gate.

Figure 14 shows a sample output of this program with 13 slices of a 48-ply graphite-epoxy test specimen. The slices represent peak-amplitude measurements from consecutive 0.25- μ s-wide gates, with the first gate set at 0.5 μ s after the front surface echo. The full-volume waveform technique interprets several hundred bytes of information at each specimen location, which leads to an extremely high storage requirement. For example, an area of 25 by 25 mm of a 32-ply graphite-epoxy composite sample scanned in steps of 0.5 mm requires approximately 5.5 megabytes of storage. This makes this method valid for in-depth analysis of particular areas but prohibits its use for inspection of large structures at this time. However, future advances in computer memory and microcomputer speed will make this technology much more affordable and applicable.

TABLE 1—Statistical comparison of subsections of 32-ply specimen.

Sample 2.32, Back Echo	Good Area (#1)	Matrix Plies (#2)	Backing Paper (#3)	Frekote Spray (#4)	Average
Total number of pixels	4160	4160	4160	4160	4160
Peak value, y _{max}	310	218	177	122	207
Peak level, $\bar{X}(y_{\max})$	117	9	74	65	66
Average level, μ	116.04	48.39	81.29	64.01	77.43
Standard deviation, δ	7.70	43.02	16.39	16.11	20.80
Skewness	-0.07	0.12	-0.08	-0.09	-0.03

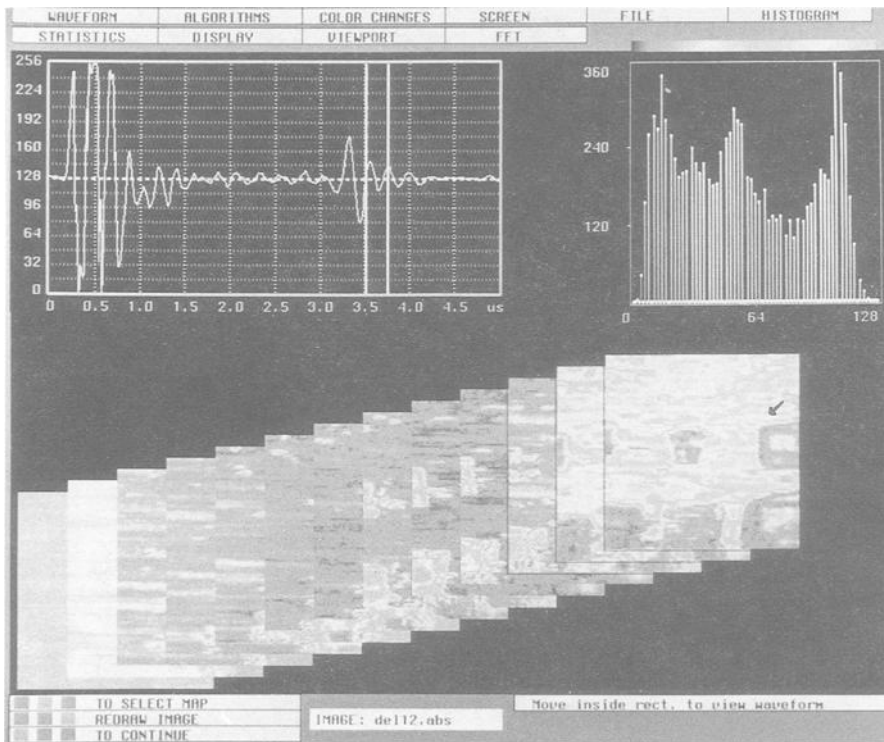


FIG. 14—Full-volume waveform analysis package with 13 slices of 48-ply graphite-epoxy specimen.

The unique capability of full-volume waveform analysis will help solve some of the problems associated with current nondestructive evaluation techniques. Up to 20 different gate settings can be selected off-line and interactively on-screen from the stored waveform database. The calculated peak amplitudes are displayed on the screen, resulting in multiple C-scans representing different depths in the sample. Gated peak detection is just one of the analysis tools implemented in this software package. Integration and standard deviation imaging techniques are available, as are simultaneous FFTs of the waveform to show its frequency spectrum. In addition, the program supports statistical procedures such as mode, median, and mean calculation of the stored waveforms [9].

Figure 15 highlights the 3-D capabilities of the program. The images provide an interface to the full ultrasonic information. In contrast to traditional scan methods, the complete information about a scanned image is available and reevaluation is possible at all times. The synthesis of digital signal- and picture-processing methods and advanced computer graphics provides a powerful means for the nondestructive evaluation of composite materials based on full-volume waveform databases.

This approach is the precursor to additional work anticipated to lead to the development of an expert system for nondestructive interrogation. The objective of this expert system is to aid the inspector in classifying abnormalities found in composite structures based on the information available through ultrasonic inspection techniques.

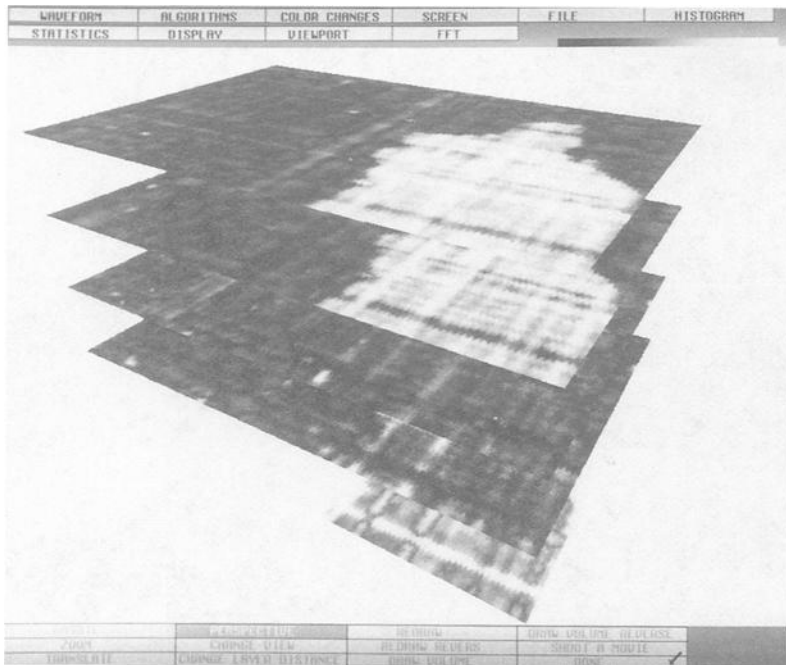


FIG. 15—Three-dimensional display of full-volume waveform software with 32-ply graphite-epoxy part with intentional Frekote spill.

Conclusions

Ultrasonic inspection methods combined with digital image enhancement techniques are a powerful tool for the classification of flaws. The introduction of a digitized full-volume waveform analysis system has enabled the inspector to establish correlations between ultrasonic waveforms and a variety of defect classes in composite structures. Statistical decision models are being developed to reduce the inspection time and the educational background required to interpret ultrasonic images.

Acknowledgments

This research was funded through the University-Industry Consortium “Applications of Composite Materials to Industrial Products” of the Center for Composite Materials (CCM) at the University of Delaware.

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Session VII

Interlaminar Fracture Analysis and Test

T. Kevin O'Brien, Chairman

Overview

There is a growing recognition in the design community of the importance of being able to calculate, test, or have the interlaminar properties of composites for use in the analysis of edge effects and critical interlaminar effects in composite structures. In addition, the designer may need and want to have knowledge of ways to suppress these effects and to evaluate the structure's damage tolerance. In this session, free edge interlaminar stresses, tensile radial stresses induced by bending, and the use of film adhesive interlayers to suppress delamination are covered.

Fish and O'Brien's "Free Edge Stress Analysis of Glass-Epoxy Laminates with Matrix Cracks," presents the results of a three-dimensional (3-D) finite element analysis to determine the interlaminar stresses near a laminate free edge in the presence of an in-plane matrix crack. They model matrix cracks in the 15° surface plies of $(15/-15)_s$ and $(15/90_n/-15)_s$ glass-epoxy laminates. Experimental results from tension tests show that delaminations initiate at the free edge in the $15/-15$ or $15/90$ interface and are bounded by the matrix crack in the 15° surface ply. The 3-D analysis indicates that large interlaminar normal tensile stresses are present in these interfaces, in addition to the interlaminar shear stresses, where the matrix crack intersects the free edge. Hence, the interlaminar normal stresses may play a significant role in the formation of delamination in these laminates.

Martin in "Delamination Failure in a Unidirectional Curved Composite Laminate" analyzes and tests graphite-epoxy unidirectional curved laminates that delaminate as a result of the tensile radial stresses induced by an applied bending moment. Initially, no stress singularities exist in these unidirectional laminates. The radial stresses have a finite distribution, which he calculates using both an anisotropic elasticity solution and a 2-D finite element analysis. Therefore, a strength-based failure criterion is used to predict the initial delamination. Then, a fracture mechanics model is used to describe the growth and accumulation through the thickness of subsequent delaminations. The progression of failure is unstable, as predicted, and delaminations continue to accumulate through the thickness until the curved laminate loses all its stiffness.

In Lagace and Bhat's "Efficient Use of Film Adhesive Interlayers to Suppress Delamination," film adhesive interlayers are placed in $(\pm 15_7/0_7)_s$ graphite epoxy laminates to suppress the initiation of delamination. The adhesive interlayer is placed only at the critical $15/-15$ interface where delamination was observed. Several different width strips are used to ascertain the effect of strip width on the ability to suppress delamination onset. For all the specimens with film adhesive interlayers, the load at which delaminations initiate is 50% higher than the control specimens without adhesive strips, and they fail at loads 40% higher than the control specimens. These improvements are obtained for all the laminates, regardless of the strip width.

There is a growing recognition in the design community of the importance of being able to calculate, test, or have the interlaminar properties of composites for use in the analysis of edge effects and critical interlaminar effects in composite structures. In addition, the designer may need and want to have knowledge of ways to suppress these effects and to evaluate the structure's damage tolerance.

Free-Edge Stress Analysis of Glass-Epoxy Laminates with Matrix Cracks

REFERENCE: Fish, J. C. and O'Brien, T. K., "Free-Edge Stress Analysis of Glass-Epoxy Laminates with Matrix Cracks," *Composite Materials: Testing and Design (Tenth Volume)*, ASTM STP 1120, Glenn C. Grimes, Ed., American Society for Testing and Materials, Philadelphia, 1992, pp. 348–364

ABSTRACT: An experimental and analytical study was conducted on the influence of matrix cracking on delamination in $(+15/90_n/-15)_s$ glass-epoxy laminates subjected to monotonically increasing tension loads. Four different layups are examined with the number of symmetric 90° layers varying from 0 to 3 ($n = 0, 1, 2, 3$). Experimental results show that local delaminations form at the intersection of matrix cracks in the $+15^\circ$ plies and the free edge. A quasi-three-dimensional (Q3D) finite-element analysis is used to examine free-edge stresses in undamaged laminates. Q3D results show that (1) in-plane transverse tensile stresses exist in the $+15^\circ$ plies near the free edges of all of the laminates and (2) only the interlaminar shear stress is high at the $+15/0$ interface. Three-dimensional (3D) finite-element analyses are used to model a $+15^\circ$ ply crack near the laminate free edge. The 3D analysis results indicate that large tensile interlaminar normal, as well as shear, stresses develop at the intersection of the matrix crack and the free edge. This suggests that the interlaminar normal stress plays a significant role in the failure of these laminates and emphasizes the importance of modeling matrix cracking in delamination analyses of composite structures.

KEYWORDS: composite materials, laminates, delamination, interlaminar fracture, matrix ply cracking, three-dimensional finite-element analysis, free-edge effect, glass-epoxy

Nomenclature

E	Young's modulus
G	Shear modulus
H	Average ply thickness
l	Matrix crack length
n	Number of 90° layers
X	Coordinate axis along laminate length
Y	Coordinate axis across laminate width
Z	Coordinate axis through laminate thickness
θ	Lamination angle
η_{xyx}	Mutual influence coefficient of the second kind
ν	Poisson's ratio
$\bar{\sigma}_{xx}$	Average applied axial stress
σ_o	Average applied axial stress at delamination onset

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Introduction

The failure of composite laminates is typically progressive, beginning with matrix cracking, followed by delamination and, finally, fiber failure. The mechanisms involved in the development of delamination are of particular interest because the stiffness reduction associated with delamination often limits the useful life of composite structures. This is especially true in dynamic systems such as helicopter rotors, where small changes in structural stiffness can adversely affect the operation of the aircraft.

Delamination has been observed to initiate near laminate free edges in the presence of matrix cracks [1–5]. Edge delamination may occur if there is a large Poisson mismatch among the plies in a laminate. This is characterized by strip delaminations along the free edges that grow into the laminate width [6]. Delaminations may also form across the laminate width, initiating at the intersection of the free edge and a matrix crack. This has been referred to as local delamination [7]. Reference 2 provides a good description of the transition from one type of delamination to the other in $(\pm 25/90_n)_s$ graphite-epoxy laminates. Previous analytical studies have shown that stress-free edges and transverse cracks interact to produce large interlaminar effects [8,9].

This paper examines the effect of matrix cracks on delamination and interlaminar stresses in a family of angle-ply laminates. Testing of the laminates is conducted to observe damage development and determine delamination initiation loads. Three-dimensional (3D) and quasi-three-dimensional (Q3D) finite-element analyses are used to model the free-edge stress states in the laminates with and without a matrix crack, respectively.

Experimental Program

Test Specimens and Procedure

Experimental testing of $(+15/90_n/-15)_s$ S2/SP250 glass-epoxy lay-ups was conducted. Four different lay-ups were examined with the number of symmetric 90° layers varying from 0 to 3 ($n = 0, 1, 2, 3$). This family of lay-ups was selected because of the large difference in the engineering properties of the plies at the $+15/\theta$ interface when an axial load is applied. The Poisson's ratio, ν_{xy} , and the mutual influence coefficient of the second kind [10], $\eta_{xy,xy}$, for glass-epoxy are plotted in Fig. 1 for different lamination angles. These engineering properties represent the in-plane transverse deformation and shearing behavior, respectively, of a ply when an axial load is applied. Mismatches in these engineering properties influence delamination in composite laminates [11]. Thus, based on large differences in $\eta_{xy,xy}$, interlaminar shear would be expected to be predominant in the delamination of these laminates.

The test specimens were 25.4 mm (1 in.) wide and 254 mm (10 in.) long. All specimens were placed in a low-temperature oven at 38°C (100°F) and weighed daily until all moisture was removed. A displacement gage was mounted at the center of the test specimens to measure axial strain as load was applied quasistatically. Black ink was swabbed onto the specimen edges just before testing to enhance visual monitoring of damage development. In addition, a high-intensity light was placed behind the specimen to improve damage detection capabilities.

Results

The first evidence of damage in the $n = 1, 2, 3$ laminates occurred in the form of matrix cracks in the 90° plies. The 90° -ply cracks traversed across most of the laminate width. This was followed by matrix cracking in the 15° plies, which were localized near the laminate free edges. In general, the 15° matrix cracks were symmetric through the laminate thickness, but not across the laminate width. Subsequently, localized delaminations formed near the free

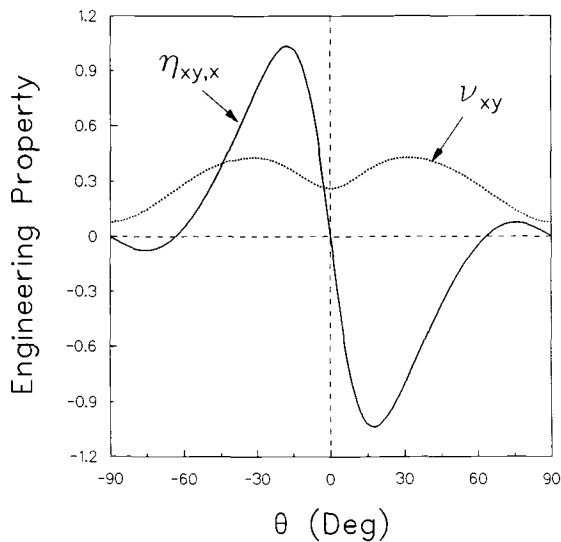


FIG. 1—Engineering properties for glass-epoxy.

edges, adjacent to the +15° matrix cracks. A photograph representative of the +15° matrix cracks and localized delaminations that develop in these laminates is shown in Fig. 2. The dark lines in the figure are tracer fibers embedded in the prepreg. The inclined white lines near the laminate free edges are +15° matrix cracks. A region surrounding the intersection of a matrix crack and the free edge is magnified in the photograph on the right (see Fig. 2). A localized delamination, which appears as a white area bounded by the matrix crack and the free edge, is shown in the magnified photograph. The subsequent damage event, which occurred shortly after the localized delaminations formed, was unstable delamination across the entire laminate width.

Damage development in the (±15), laminates was not observed before unstable delamination across the laminate width, as shown in Fig. 3. However, the test results from the $n =$

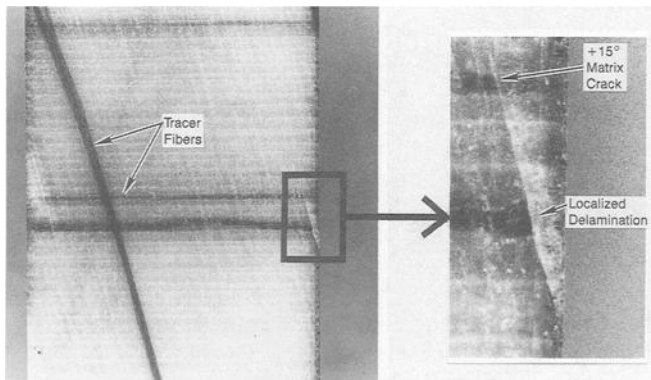


FIG. 2—Photographs of matrix cracking and localized delaminations in a (+15/90₄/−15), laminate.

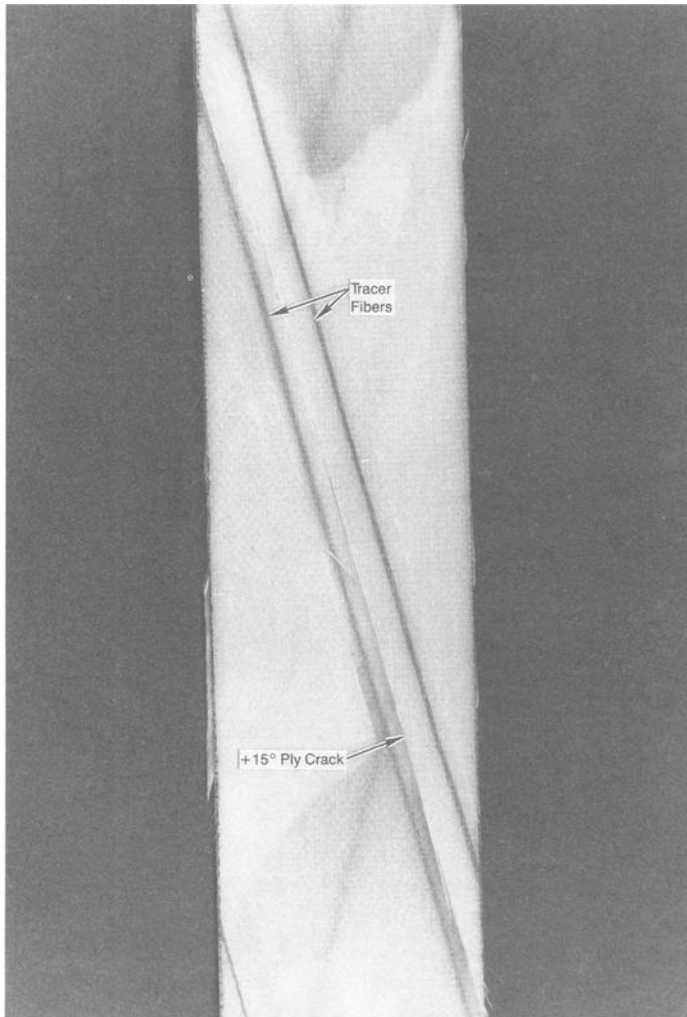


FIG. 3—Photograph of delamination in a $(\pm 15)_s$ laminate.

1, 2, and 3 lay-ups suggest that matrix cracking in the $+15^\circ$ plies also occurs in the $(\pm 15)_s$ laminates before delamination.

The average applied axial stresses, $\bar{\sigma}_{xx}$, for each of the damage events are plotted in Fig. 4 as a function of the number of 90° plies in the lay-up, n . Each data point represents the average of five test specimens. As n increases, the average applied axial stress for delamination decreases. Similarly, the average applied axial stress for the other damage events decreases as n increases from 1 to 3. The narrow margin between matrix cracking in the $+15^\circ$ plies and the onset of localized delamination is apparent in Fig. 4. Schematic illustrations of the damaged laminates just before and after delamination are given in Fig. 5. These experimental observations are the basis for the analytical modeling described below.

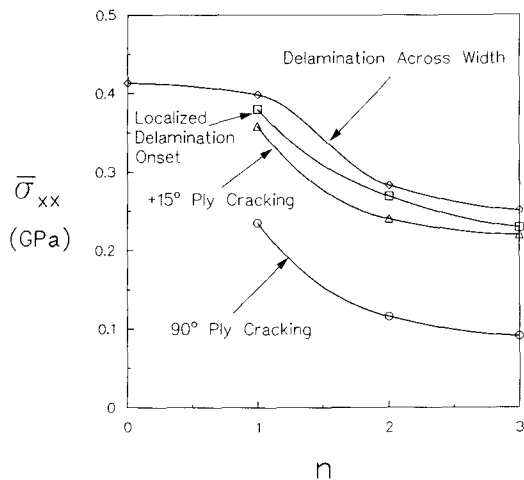


FIG. 4—Damage development in $(+15/90_n/-15)_s$ laminates.

Stress Analysis

Q3D and 3D finite-element models are used to analyze stresses in the $(+15/90_n/-15)_s$ family of laminates. Q3D analysis is used to examine free-edge stresses in the undamaged (before $+15^\circ$ matrix cracking) laminates. The 3D analysis is used to model the effect of a single matrix crack in the $+15^\circ$ ply on the free-edge stresses in the laminates.

The material properties used to model the glass-epoxy laminates are given in Table 1. The

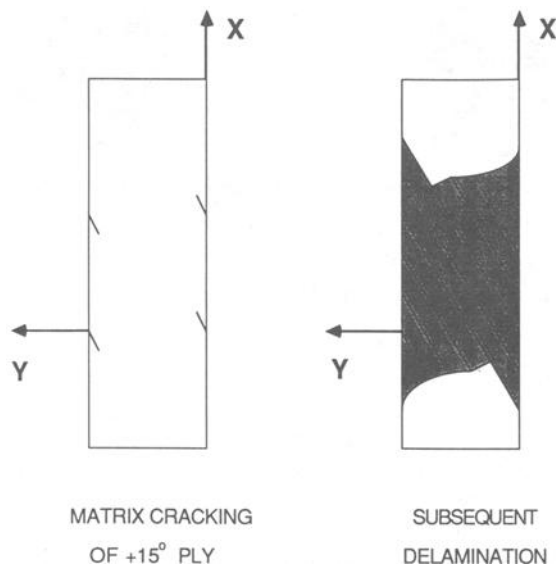


FIG. 5—Schematic illustrations of damaged laminates just before and after delamination.

TABLE 1—*S2/SP250 material properties.*

$E_1 = 43 \text{ GPa } (6.3 \times 10^6 \text{ psi}), E_2 = E_3 = 10 \text{ GPa } (1.5 \times 10^6 \text{ psi})$
$\nu_{12} = \nu_{13} = 0.26, \nu_{23} = 0.41$
$G_{12} = G_{13} = 5.5 \text{ GPa } (0.80 \times 10^6 \text{ psi}), G_{23} = 3.7 \text{ GPa } (0.53 \times 10^6 \text{ psi})$
$\alpha_1 = 4.40 \times 10^{-6} \text{ mm/mm/}^\circ\text{C } (2.44 \times 10^{-6} \text{ in./in./}^\circ\text{F})$
$\alpha_2 = 26.6 \times 10^{-6} \text{ mm/mm/}^\circ\text{C } (14.8 \times 10^{-6} \text{ in./in./}^\circ\text{F})$

subscripts 1, 2, and 3 represent the longitudinal, transverse, and thickness directions, respectively, of a unidirectional ply. The in-plane elastic properties were determined experimentally at NASA Langley Research Center. The out-of-plane Poisson’s ratio, ν_{23} , and shear modulus, G_{23} , were calculated using the measured in-plane properties and FEMA (Finite Element Micromechanics Analysis), a computer program developed at McDonnell Douglas Helicopter Co. (MDHC) [12]. The material was assumed to be transversely isotropic. The thermal expansion coefficients were obtained from the material manufacturer (3M Corp.). The measured average ply thickness for the laminates, H , was 0.2159 mm (0.0085 in.).

Throughout this analysis, a temperature differential of $-115^\circ\text{C } (-175^\circ\text{F})$ is applied to represent residual curing effects. In addition, the transverse modulus of the 90° plies is reduced to 20% of the initial value ($0.2 E_2$) to account for the laminate stiffness reduction caused by 90° ply cracking. Laminate moduli calculated using this reduced value were found to approximate the experimentally observed laminate stiffnesses just before localized delamination onset in the $n = 1, 2$, and 3 laminates to within $\pm 6\%$.

Q3D Finite-Element Models

The Q3D models represent through-the-thickness cross sections ($x = \text{constant}$) of the laminates over which a state of generalized plain strain exists [10]. The Q3D mesh for the $(\pm 15)_s$ laminate is shown in Fig. 6. A fine mesh adjacent to the free edge ($Y = 0$) consisting of 20 elements one quarter of a ply thickness wide (Y direction) is used to examine the free-edge stress state. Each of the plies in the $(\pm 15)_s$ laminate is modeled with three elements through the thickness (z direction). Additional layers of elements of one ply thickness are added adjacent to the laminate mid-plane in the models for the $n = 1, 2$, and 3 laminates. For example, the finite element mesh for the $(+15/90_3/-15)_s$ laminate consists of nine layers of elements (six with a thickness of $H/3$ and three with a thickness of H).

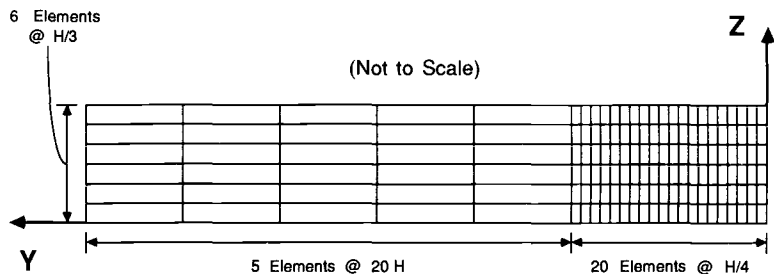
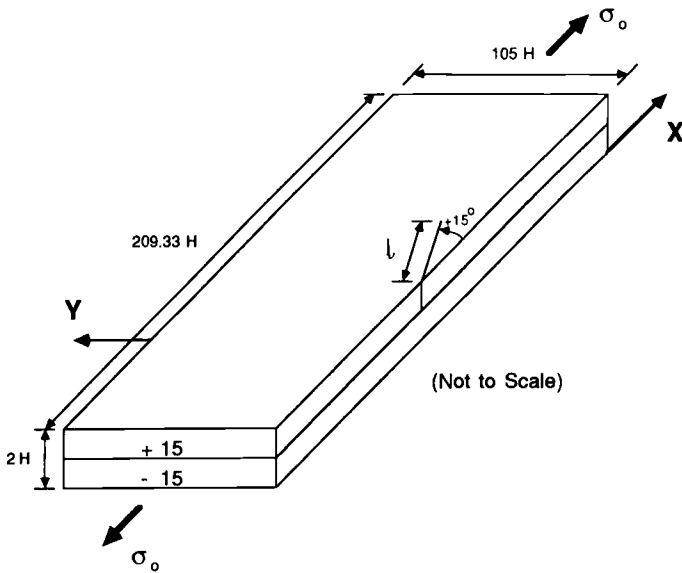


FIG. 6 Q3D mesh for $(\pm 15)_s$ laminate.

FIG. 7—Modeled region of $(\pm 15)_s$ laminate.

3D Finite-Element Models

A portion of the laminates with a matrix crack in the $+15^\circ$ ply intersecting the free edge is modeled with 3D finite elements. The 3D model for the $(\pm 15)_s$ laminate is shown in Fig. 7, where l is the matrix crack length. The applied axial stress, σ_o , is also shown in Fig. 7. The model is approximately 209 ply thicknesses in length (X direction) and 105 ply thicknesses in width (Y direction). Symmetry conditions are imposed so that only one half of the laminate thickness needs to be modeled. Eight-noded hexahedral elements are used along with six-noded pentahedral elements to discretize the region. The hexahedral elements are based on an assumed stress hybrid formulation [13]; the pentahedral elements are isoparametric.

The finite element mesh used for the $(\pm 15)_s$ laminate is shown in Fig. 8. A fine mesh in planes normal to the Y coordinate direction is shown by the shaded region Fig. 8. The discretization is identical to the fine mesh adjacent to the free edge in the Q3D model in Fig. 6. As with the Q3D models, three layers of elements are used to model each of the plies adjacent to the $+15/\theta$ interface. Additional layers of elements of one ply thickness are added adjacent to the laminate mid-plane for the $n = 1, 2$, and 3 laminates.

A top view of a portion of the mesh surrounding the matrix crack is given in Fig. 9. The pentahedral elements are used to model the matrix crack and appear as triangles in Fig. 9. Three layers of pentahedral elements are used to model a matrix crack through the entire thickness of the $+15^\circ$ ply.

Q3D Analysis Results

The MDHC Q3D finite-element code, FRESA (*F*ree-Edge Stress Analysis), was used to calculate free-edge stresses in the undamaged laminates. All stress results are normalized by σ_o , which represents the average applied axial stress at which delamination across the laminate width was experimentally observed. The normalized in-plane transverse stress, σ_{22} , in the mid-

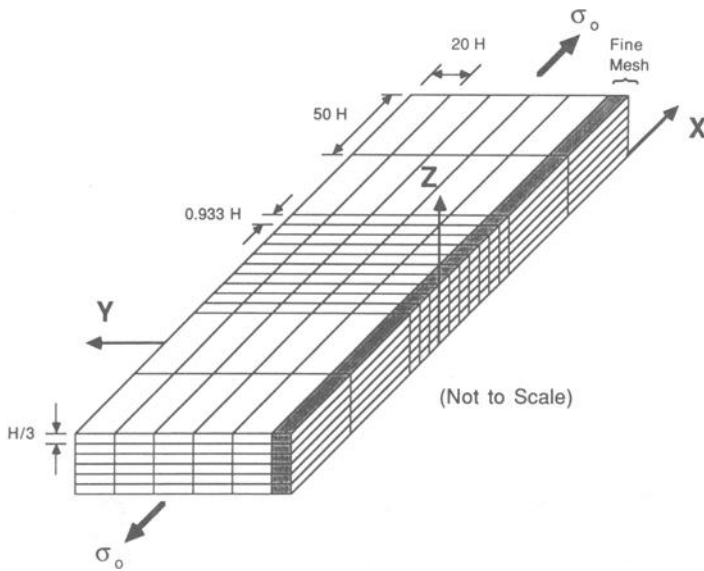


FIG. 8—Finite element mesh for $(\pm 15)_s$ laminate.

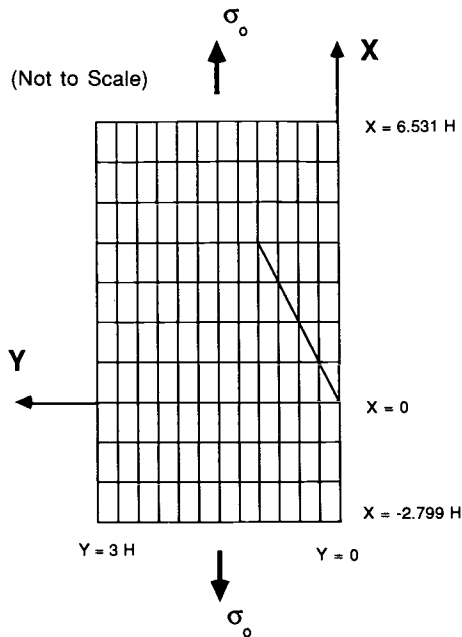


FIG. 9—Top view of portion of mesh surrounding matrix crack.

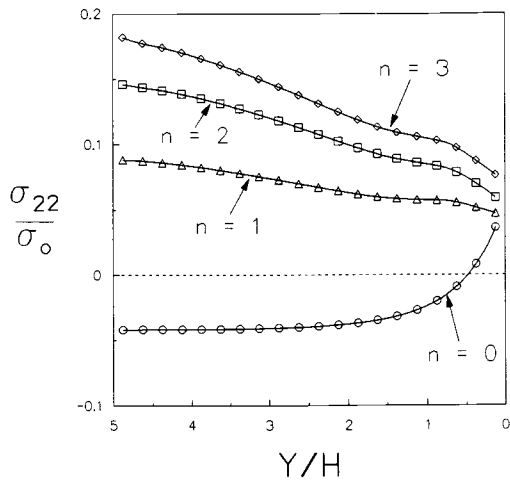


FIG. 10—In-plane transverse stresses in +15° plies near free edge using Q3D analysis.

dle of the +15° ply near the free edges of the laminates are shown in Fig. 10. The plotted data points represent average values for the layer of elements in the middle of the +15° ply. In the laminate interior, the in-plane transverse stresses for the $n = 1, 2$, and 3 laminates are all tensile and decrease in magnitude as the free edge ($Y/H = 0$) is approached. However, the transverse stress for the $(\pm 15)_s$ laminate is compressive in the interior of the laminate and becomes tensile only near the free edge. Thus, the Q3D results in Fig. 10 show that in-plane transverse tensile stresses, which promote matrix cracking in the +15° plies, exist near the free edge in all of the laminates.

The Q3D interlaminar stresses along the +15/−15 interface near the free edge of the $(\pm 15)_s$ laminate are given in Fig. 11. The plotted data points represent average values for the elements across the +15/−15 interface. The interlaminar shear stress, σ_{xz} , has the largest mag-

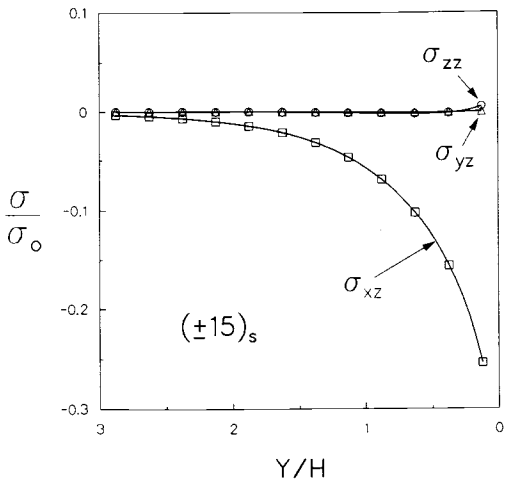


FIG. 11—Interlaminar stresses in undamaged $(\pm 15)_s$ laminate along +15/−15 interface using Q3D

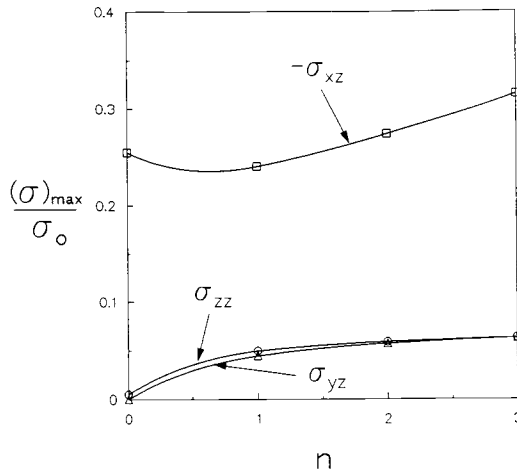


FIG. 12—Peak interlaminar stresses at the $+15/\theta$ interface in $(+15/90_n/-15)_s$ laminates from Q3D analysis.

nitude at the free edge. The interlaminar normal stress, σ_{zz} , and the interlaminar shear stress, σ_{yz} , are essentially zero. The peak magnitudes of the Q3D interlaminar stresses at the $+15/\theta$ interface in all of the layups considered are shown in Fig. 12. In general, as the number of 90° plies is increased, the magnitudes of the interlaminar stresses also increase. The only exception is a slight decrease in σ_{xz} when n increases from 0 to 1. The interlaminar stress states in the $n = 1, 2$, and 3 laminates continue to be dominated by σ_{xz} , although the relative contributions of the remaining two stress components have increased relative to the $n = 0$ laminate.

3D Analysis Results

The 3D model is now used to examine the effects of a $+15^\circ$ -ply crack on the free-edge stress state in the $(\pm 15)_s$ laminate. The in-plane transverse stress in the $+15^\circ$ ply surrounding the matrix crack near the laminate free edge is shown in Fig. 13. Results for four different matrix crack lengths ($l/H = 0.966, 1.932, 2.898$, and 3.864) are presented in Fig. 13. The region over which the stresses are plotted is represented in Fig. 9. As the matrix crack grows away from the free edge, the magnitude of the peak transverse stress continues to increase. This suggests that unstable matrix cracking may occur in this laminate and is corroborated by the experimental observations.

The interlaminar normal stress for the four different matrix crack lengths in the $(\pm 15)_s$ laminate is shown in Fig. 14. When the matrix crack is small ($l/H = 0.966$), σ_{zz} is essentially zero, except where the matrix crack meets the free edge. However, σ_{zz} increases markedly as the matrix crack grows, and the area over which the increased σ_{zz} acts is also expanded. The results for the interlaminar shear stress, σ_{yz} , are given in Fig. 15. The results show that σ_{yz} is relatively insensitive to the presence and growth of a matrix crack. Finally, the effect of matrix crack growth on σ_{xz} is shown in Fig. 16. As with the results for σ_{22} and σ_{zz} , σ_{xz} increases significantly as the matrix crack grows away from the free edge.

The σ_{zz} and σ_{xz} stress distributions along the free edge of the $(\pm 15)_s$ laminate are given in Figs. 17 and 18, respectively. The stresses are averaged across the elements nearest the free edge ($Y = 0.125H$) at the $+15/-15$ interface ($Z = H$). Both figures show that the stresses along the free edge increase rapidly as the matrix crack grows. The magnitude of σ_{zz} relative to σ_{xz}

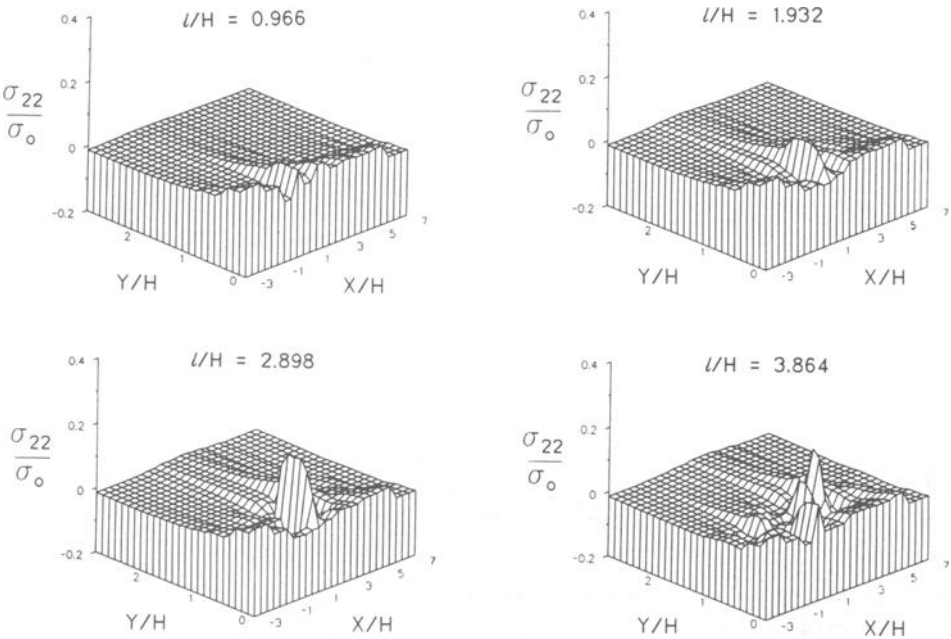


FIG. 13—In-plane transverse stress surrounding matrix cracks of different lengths in the $+15^\circ$ ply of the $(\pm 15)_S$ laminate.

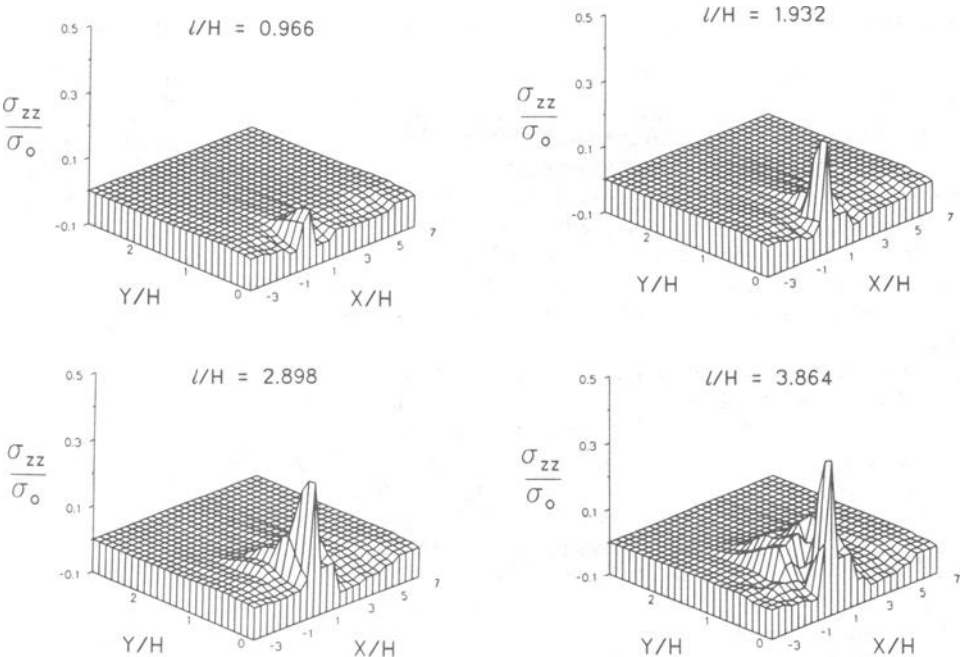


FIG. 14— σ_{zz} distribution surrounding matrix cracks of different lengths at the $+15/-15$ interface of the $(\pm 15)_S$ laminate.

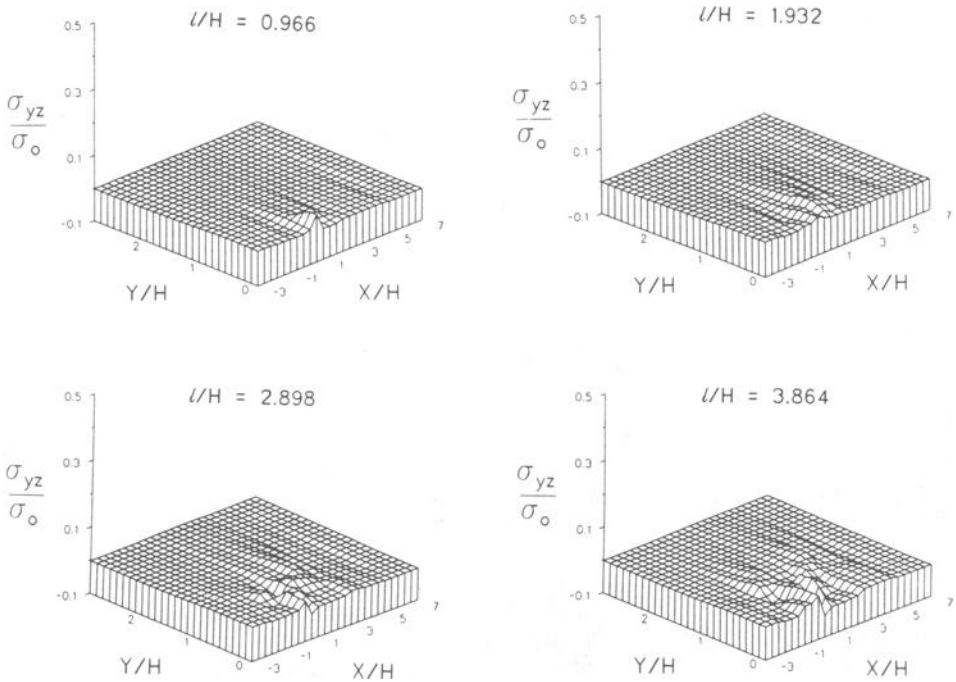


FIG. 15— σ_{yz} distribution surrounding matrix cracks of different lengths at the $+15/-15$ interface of the $(\pm 15)_s$ laminate.

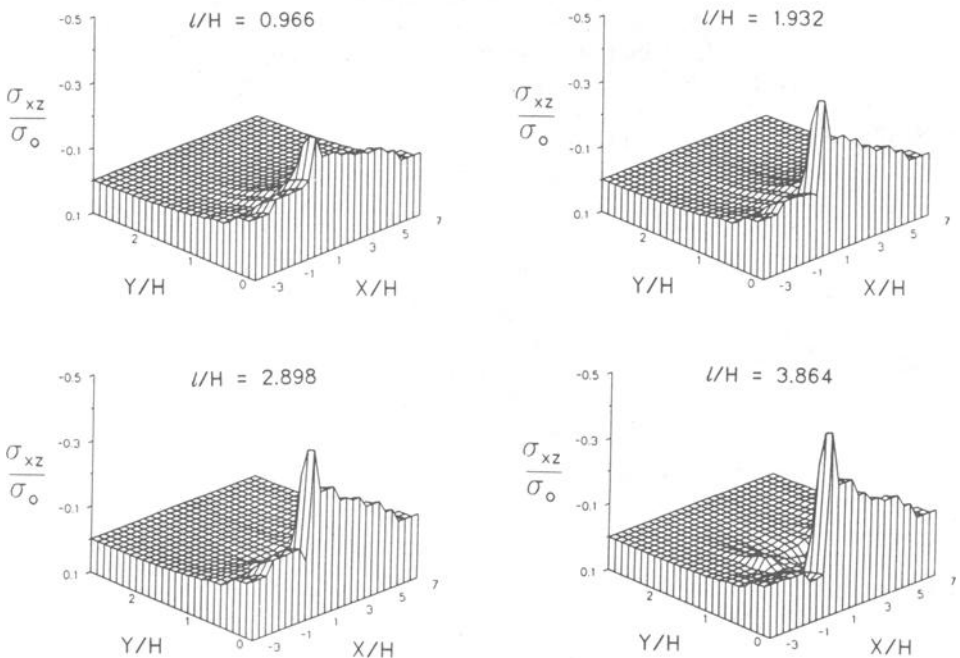


FIG. 16—The σ_{xz} distribution surrounding matrix cracks of different lengths at the $+15/-15$ interface of the $(\pm 15)_s$ laminate.

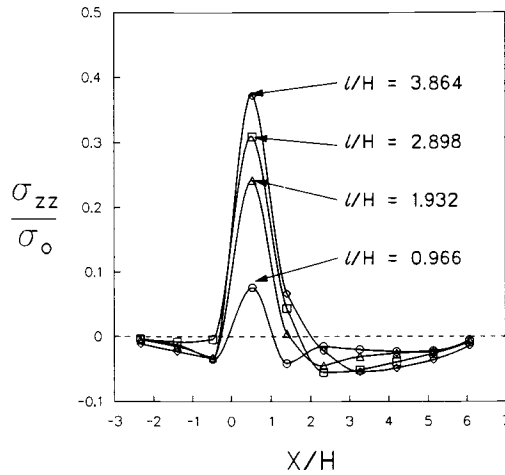


FIG. 17—Effect of matrix crack length on σ_{zz} distribution near the free edge of the $(\pm 15)_s$ laminate from 3D analysis.

increases from 0.29 to 0.81 as the matrix crack grows from $l/H = 0.966$ to $l/H = 3.864$. Furthermore, the peaks in the stress distributions occur just ahead of where the matrix crack meets the free edge (that is, $0 < X/H < 1$), indicating that this area is the most prone to delamination. In fact, the photograph of the localized delamination in Fig. 2 substantiates this deduction. All observed localized delaminations formed in this area and were bounded by the matrix crack and the free edge.

Free-edge stress results for the $n = 1, 2$, and 3 laminates show some differences compared to the results for the $(\pm 15)_s$ laminate. The effect of the matrix crack length on the transverse stress in the $(+15/90/-15)_s$ laminate is shown in Fig. 19. As before, the stresses are in the middle of the $+15^\circ$ ply. The transverse stress distribution in the $+15^\circ$ ply of the $(+15/90/-15)_s$ laminate is less affected by matrix crack growth than the $(\pm 15)_s$ laminate is (see Fig. 13). Similar results were found for the $n = 2$ and $n = 3$ laminates. This indicates that, com-

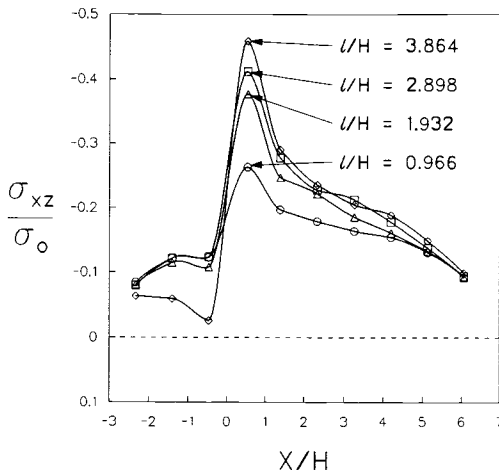


FIG. 18—Effect of matrix crack length on σ_{xz} distribution near the free edge of the $(\pm 15)_s$ laminate from 3D analysis.

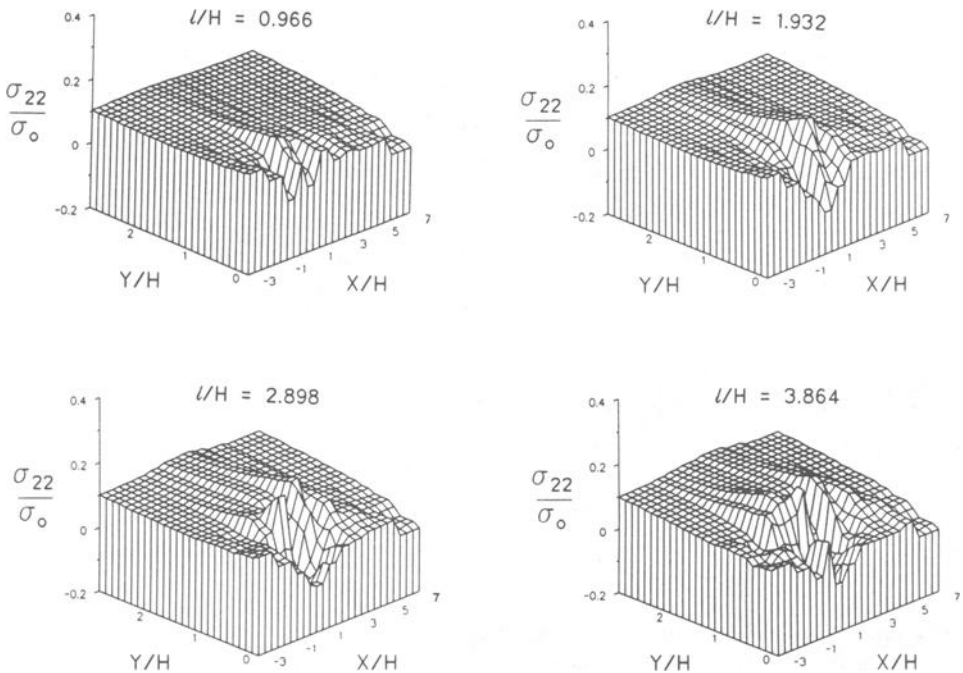


FIG. 19—In-plane transverse stresses surrounding matrix cracks of different lengths in the $+15^\circ$ ply of the $(+15/90/-15)_3$ laminate.

pared to the $n = 0$ laminate, the $n = 1, 2$, and 3 laminates may not be as prone to unstable matrix cracking in the $+15^\circ$ plies. The experimental observations support this inference.

The effect of the matrix crack length on the peak magnitudes of σ_{zz} and σ_{xz} ($[\sigma_{zz}]_{\max}$ and $[\sigma_{xz}]_{\max}$) for all of the laminates is shown in Figs. 20 and 21, respectively. The peak magnitudes

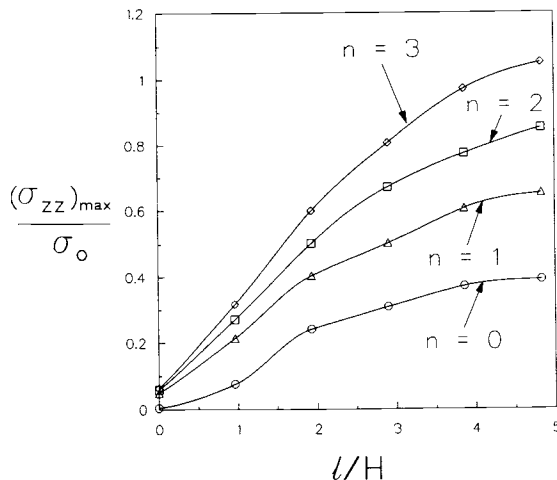


FIG. 20—Effect of matrix crack length on the peak interlaminar normal stress at the $+15/\theta$ interface in the $(+15/90/-15)_3$ laminate.

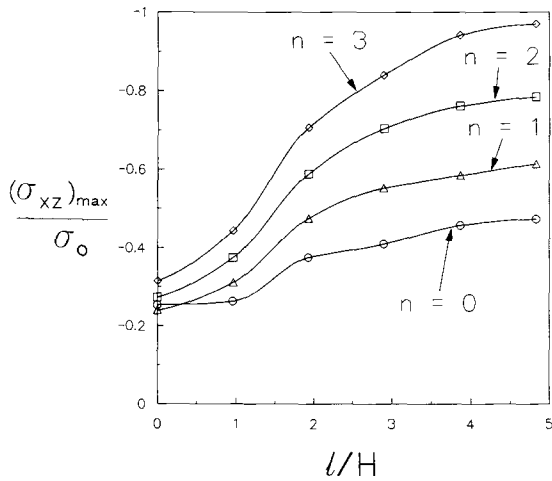


FIG. 21—Effect of matrix crack length on the peak interlaminar shear stresses at the +15/θ interface in (+15/90_n/−15)_s laminates.

of σ_{zz} and σ_{xz} for all of the laminates occur along the free edge at the +15/θ interface adjacent to the matrix crack ($0 < X/H < 1$, as in Figs. 17 and 18). Both $(\sigma_{zz})_{\max}$ and $(\sigma_{xz})_{\max}$ increase as the matrix crack length increases and appear to be approaching limiting values for each of the laminates at matrix crack lengths slightly greater than $l/H = 5$. In addition, the magnitudes of $(\sigma_{zz})_{\max}$ relative to $(\sigma_{xz})_{\max}$ for the laminates increase as the matrix crack length increases (Fig. 22). In fact, at a matrix crack length of $l/H = 4.830$, $(\sigma_{zz})_{\max}$ for the $n = 1, 2$, and 3 laminates exceeds $(\sigma_{xz})_{\max}$.

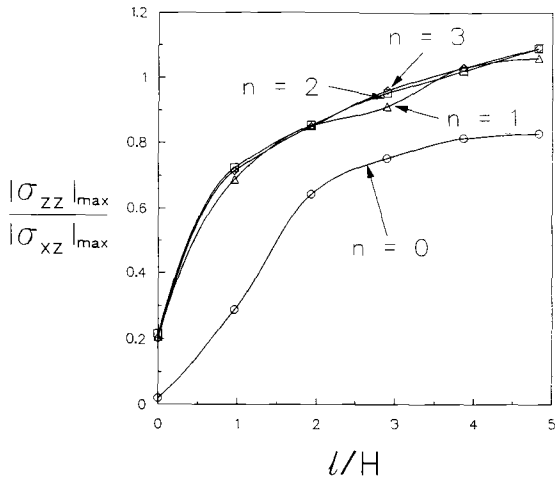


FIG. 22—Effect of matrix crack length on the peak magnitudes of σ_{zz} relative to σ_{xz} at the +15/θ interface in (+15/90_n/−15)_s laminates.

Conclusions

An experimental and analytical study was conducted on the influence of matrix cracking on free-edge stresses in $(+15/90_n/-15)_s$ S2/SP250 glass-epoxy laminates. Four different lay-ups, with the number of symmetric 90° layers varying from 0 to 3 ($n = 0, 1, 2, 3$), were subjected to monotonically increasing tensile loads. A Q3D finite-element analysis was used to examine free-edge stresses in laminates without $+15^\circ$ ply cracks. A 3D finite-element analysis was used to investigate stresses caused by a $+15^\circ$ ply crack near the laminate free edge.

The experimental results showed that

- 1 As n increased from 0 to 3, the average applied axial stress for delamination decreased
- 2 Localized delaminations in the laminates were bounded by $+15^\circ$ ply cracks and the laminate free edges
- 3 A narrow margin between localized delamination onset and unstable delamination across the laminate width was observed for the $n = 1, 2$, and 3 laminates

The Q3D analysis results for laminates without a $+15^\circ$ ply crack showed that

- 1 In-plane transverse tensile stresses, which promote matrix cracking, existed in the $+15^\circ$ plies near the free edges of all of the laminates
- 2 The interlaminar shear stress, σ_{xz} , was significantly higher than the interlaminar normal stress, σ_{zz} , at the $+15/\theta$ interface of the laminates

The 3D analysis results for laminates with a $+15^\circ$ ply crack near the free edge showed that

- 1 Large tensile interlaminar normal, as well as shear (σ_{xz}), stresses existed at the $+15/\theta$ interface of the laminates
- 2 The interlaminar stress magnitudes were greatest near the intersection of the matrix crack and the free edge
- 3 The interlaminar stresses increased as the matrix crack grew away from the laminate free edge and appeared to be approaching limiting values at matrix crack lengths slightly greater than five ply thicknesses
- 4 The magnitude of the interlaminar normal stress relative to the interlaminar shear stress, σ_{xz} , increases rapidly as the matrix crack grows away from the free edge
- 5 The magnitudes of the interlaminar stress components, σ_{zz} and σ_{xz} , increased as the number of symmetric 90° plies in the laminate increased

This study shows that significant differences in interlaminar stress states (absolute and relative magnitudes) arise depending on whether matrix cracking is modeled in the analysis of free edges in composite laminates. Q3D analysis may be useful in assessing the point at which initial damage (that is, matrix cracking) occurs. However, to examine subsequent failure (that is, delamination) of the structure, an analysis that accounts for matrix cracking is required.

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Delamination Failure in a Unidirectional Curved Composite Laminate

REFERENCE: Martin, R. H., "Delamination Failure in a Unidirectional Curved Composite Laminate," *Composite Materials Testing and Design (Tenth Volume)*, ASTM STP 1120, Glenn C. Grimes, Ed., American Society for Testing and Materials, Philadelphia, 1992, pp. 365–383

ABSTRACT: Delamination failure in a unidirectional curved composite laminate was investigated. The curved laminate failed unstably by delaminations developing around the curved region of the laminate at different depths through the thickness until virtually all bending stiffness was lost. Delamination was assumed to initiate at the location of the highest radial stress in the curved region. A closed form curved beam elasticity solution and a two-dimensional finite element analysis (FEA) were conducted to determine this location. The variation in the strain energy release rate, G , with delamination growth was determined using the FEA. A strength-based failure criterion adequately predicted the interlaminar tension failure that caused initial delamination onset. Using the G analysis, the delamination was predicted to extend into the arm and leg of the laminate, predominantly in Mode I. As the initial delamination grew around the curved region, the maximum radial stress in the newly formed inner sublaminates increased to a level sufficient to cause a new delamination to initiate in the sublaminates with no increase in applied load. This failure progression was observed experimentally.

KEYWORDS: composite materials, curved laminates, delamination, fracture toughness, strain energy release rate

Nomenclature

a	Delamination length in the counter-clockwise direction
b	Delamination length in the clockwise direction
c, R_{inner}	Inner radius
d	Outer radius
E_r, E_{22}, E_{33}	Radial/transverse tensile moduli
E_θ, E_{11}	Tangential/longitudinal tensile moduli
F_r	Force in the radial direction
F_θ	Force in the tangential direction
FAW	Fiber areal weight
FD	Fiber density
G_{Ic}	Critical value of Mode I strain energy release rate
G^{SE}	Total strain energy release rate calculated from change in strain energy
G^{VT}	Total strain energy release rate calculated from virtual crack closure technique (VCCT)
G_I^{VT}	Mode I strain energy release rate calculated from VCCT
G_{II}^{VT}	Mode II strain energy release rate calculated from VCCT
$G_{\theta\theta}, G_{12}, G_{13}$	Transverse shear modulus

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h	Ply thickness
l	Total delamination length
L, L_1	Length of moment arm
M	Moment
N	Number of plies
P	End load
P_f	Load to onset of first delamination
r	Radius
R	Radial distance of delamination onset
t	Thickness of specimen
U	Complementary strain energy
V_f	Volume fraction
w	Width of beam
α	Angle of delamination in counter-clockwise direction
β	Angle of delamination in clockwise direction
δ_r	Displacement in the radial direction
δ_θ	Displacement in the tangential direction
θ	Angle around curved region from leg
κ	Anisotropic parameter for moments
λ	Length of finite element at delamination front
$\nu_{\theta\theta}, \nu_{21}$	Poisson's ratio
σ_{2f}, σ_{3f}	Transverse tensile strengths
σ_r	Radial stress
σ_r^M	Radial stress caused by moments
σ_r^P	Radial stress caused by end loads
σ_θ	Tangential stress
σ_θ^M	Tangential stress caused by moments
σ_θ^P	Tangential stress caused by end loads
τ_{12f}, τ_{13f}	Shear strengths
$\tau_{r\theta}$	r - θ shear stress
$\tau_{r\theta}^M$	r - θ shear stress caused by moments
$\tau_{r\theta}^P$	r - θ shear stress caused by end loads
ϕ	Total angle of delamination = $\alpha + \beta$
ω	Anisotropic parameter for end loads

Introduction

With the increased use of laminated fiber-reinforced composite materials in primary aircraft structural components, the ability to understand and predict their failure modes becomes paramount. Because of the low through-the-thickness strengths of composites, one of their prime modes of failure has been delamination [1]. The high interlaminar stresses that may cause delamination can arise from material discontinuities, geometric discontinuities, or eccentricities in the load path. One example of a generic structural component that has the possibility of all three of these interlaminar stress sources is the curved laminate. Some typical uses of a curved composite laminate are shown in Fig. 1. Mostly, these laminates are of a multidirectional layup. The material property mismatch between adjacent plies of different orientation may create singular interlaminar stresses at the free edge causing edge delamination [2]. Another failure mode occurs when the membrane stresses in off-axis plies are high and cause transverse tension cracking across the width [3]. Mathematically singular interlaminar stresses may occur where the matrix crack meets the adjacent plies. These stresses may cause delam-

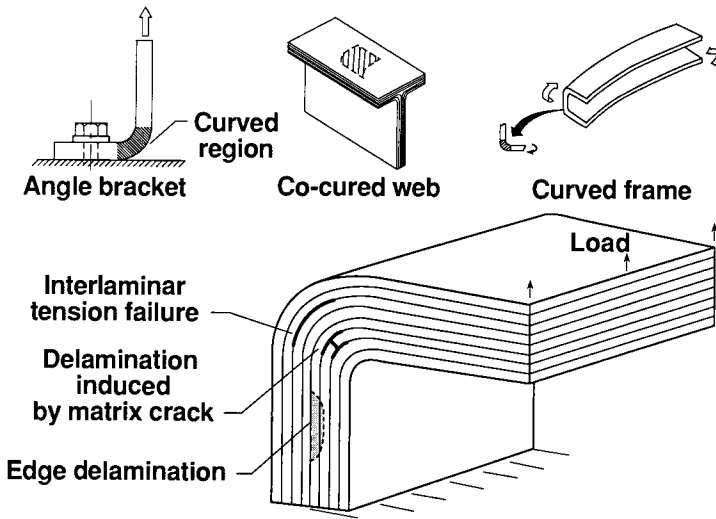


FIG. 1—Curved laminate configurations and failures modes.

ination at these locations. Delamination may also occur as an interlaminar tensile failure caused by tensile radial stresses created from the bending of the curved laminate.

Several studies in the literature have focused on determining damage in curved laminates. Finite element analysis (FEA) was used to determine the strain distribution for the design of curved composite frames for the UH60 Black Hawk Helicopter [4]. A strength-based criterion, called the Chang-Springer criterion, which was based on the Tsai-Hill criterion, was used to predict if an off-axis ply failure or an interlaminar tensile failure would occur in multidirectional curved laminates [5]. Finite element analysis and a closed-form solution were conducted to determine the stress distribution in a curved composite frame taken from the V22 Osprey [6]. The tensor polynomial criterion was used to predict failure. An elasticity solution was used to determine the stress distribution in a semicircular composite beam [7]. In Ref 8, composite angles of different lay-ups were tested and analyzed. Different failure modes were noted. In one layup, the failure was delamination from an interlaminar tension failure, and in another, failure was matrix cracking in a 90° ply followed by delamination initiating from the matrix crack. A strength-based failure criterion predicted the different failure modes but yielded generally conservative predictions.

A review of some of the work conducted on curved composite beams and frames [9] notes that predicting delamination is difficult using strength-based failure criteria because standard test methods for determining the through-the-thickness strengths, σ_{3f} and τ_{13f} , do not exist. A further problem with the use of strength-based criteria in the above examples is the determination of the stresses. Although a stress analysis may be useful to identify areas that are highly stressed, a strength-based failure criteria may only be used when no mathematically singular stresses are present. As soon as damage or a discontinuity is present, such as a radial delamination, a matrix crack, or a free edge, stresses cannot be evaluated at the singularities, and a strength criterion should not be used.

Fracture mechanics-based failure criteria offer a technique to predict the strength and life of a component with a singularity caused by a discontinuity. An attempt to use a fracture mechanics approach to predict the static strength of a curved laminate containing both a

matrix crack and a radial delamination was reported in Ref 10. The failure loads predicted were generally unconservative, possibly because the two-dimensional FEA did not account for the effects of the free edge. In Ref 11, delamination was assumed to exist in a production part in a curved solid laminate segment of a fuel sponson. The damage was assumed to occur during manufacture as a result of residual thermal stresses. A 2-D coupled boundary element fracture mechanics approach was used to determine how large the delamination could become before the component completely delaminated along that path at a given design load. It was determined that the delamination could be easily detected by nondestructive inspection (NDI) methods long before total failure of the component. A fracture mechanics-based method for predicting the static strength and the fatigue life of other structural components that experience delamination initiating from singular stress sources has been proposed [12-15]. This method uses analyses to determine the critical value of strain energy release rate, G , and its components along with delamination onset data obtained from generic test coupons [16-19].

The purpose of this paper is to predict the maximum load a unidirectional curved laminate can sustain without damage, and, once damage occurs, how it accumulates. Unidirectional laminates were used to isolate interlaminar tension failure, that is, it was ensured that no edge delamination or delamination from cracked plies would occur. Simple right-angled coupons were analyzed and tested to represent the loading on the angle bracket configuration shown in Fig. 1. First, the location of initial failure was determined using a stress analysis assuming an undamaged component containing no singular stress sources. A closed form curved beam elasticity solution and FEA were used to compute the interlaminar stresses. Second, a fracture mechanics analysis was conducted to determine damage progression after the onset of delamination. This characterization was done by determining how G varies with delamination growth. A method for computing the variation of G for an embedded delamination with two delamination fronts was developed. Further damage progression was determined by analyzing the curved laminate with the delamination present.

Materials

Twenty-four-ply, graphite/epoxy (AS4/3501-6) unidirectional curved laminates were manufactured at NASA-Langley Research Center. The laminates were cured according to the material manufacturer's instructions. The curved laminates were layed up over a solid aluminum tool in panels 300 mm wide to give a 90° bend. After cure, the panels were cut into 25-mm-wide specimens. The final dimensions of the curved laminates are given in Fig. 2. Double cantilever beam (DCB) specimens were also manufactured. The DCB specimens were 24-ply unidirectional coupons with a sheet of 0.127-mm (0.0005-in.) Kapton film placed between the twelfth and thirteenth plies before curing to simulate a delamination. The dimensions and testing procedure of the DCB are given in Ref 17. The following values of fracture toughness were obtained from delamination growth initiating at the insert of four specimens: $69.0 \leq G_{Ic} \leq 99.0 \text{ J/m}^2$ (mean = 80.0 J/m^2). Values for transverse tensile strength obtained from flat specimens were $36.5 \leq \sigma_{2T} \leq 69.0 \text{ MPa}$ (mean = 53.8 MPa) [20]. Specimens to determine the elastic properties of the composite were also manufactured. The elastic tensile and shear properties used in the FEA and the closed form elasticity solution were obtained using ASTM Test for Tensile Properties of Fiber-Resin Composites (D 3039) and ASTM Practice for In-plane Shear Stress-Strain Response of Unidirectional Reinforced Plastics (D 3518), respectively. The following average properties from at least three tests were obtained assuming transverse isotropy:

$$E_1 = E_2 = E_3 = E_{11} = E_{22} = E_{33} = E_r = 10.0 \text{ GPa}$$

$$G_{12} = G_{13} = G_{23} = 5.24 \text{ GPa}$$

$$\nu_{12} = \nu_{13} = \nu_{23} = \nu_{21} = \nu_{31} = \nu_{32} = 0.3$$

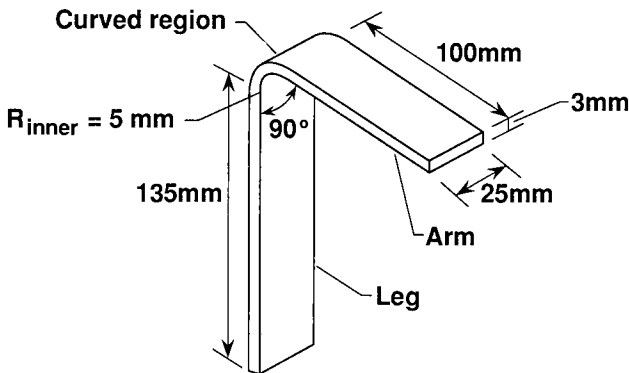


FIG. 2—Dimensions of curved laminate test specimens.

The average volume fraction, V_f , of the specimens was determined from

$$V_f = \frac{FAW \times N \times 100}{FD \times t}$$

where

FAW = fiber areal weight,

N = number of plies,

FD = fiber density, and

t = thickness of the specimen.

For the DCB specimen, V_f was determined to be 55.9%. In the curved laminate specimens, V_f in the curved region was 54.7%, and in the leg V_f was 56.1%. In the specimens used to obtain elastic moduli, V_f = 59.1%.

Experimental Procedure and Results

A fixture for testing the curved laminates in a standard tension testing machine is shown in Fig. 3. An 890-N (200-lb) load cell was incorporated into the test fixture to monitor the small loads associated with matrix failures in polymeric composites. Load was applied to the specimen at a distance L of 50 mm (Fig. 4) from the curved region/arm intersection via a hinge clamped to the specimen. The other end of the hinge was held in a clamp that was free to pivot. The displacements were small during the test because of the high stiffness of the unidirectional curved laminate. Therefore, the load applied via the hinge was considered to remain vertical during testing. A displacement transducer, not shown in Fig. 3, was attached to the side of the loading arrangement to measure vertical displacements during the tests. The edges of the specimens were painted with brittle white paint to aid visualization of delamination. The load and displacement were monitored on an x - y plotter to determine the failure load. The laminates were loaded quasistatically in load control.

Maximum static loads per unit width for damage initiation in the unidirectional curved laminates are shown in Table 1. Significant scatter can be seen in the results. Similar scatter was reported in Ref 6 and is consistent with the large scatter in transverse tensile strength reported in Ref 20. The specimens were labeled sequentially from the location they were cut from the panels. No trend was detected between the strength and the location where the curved laminate was cut from the panel. The normalized radius location of the first delamination is

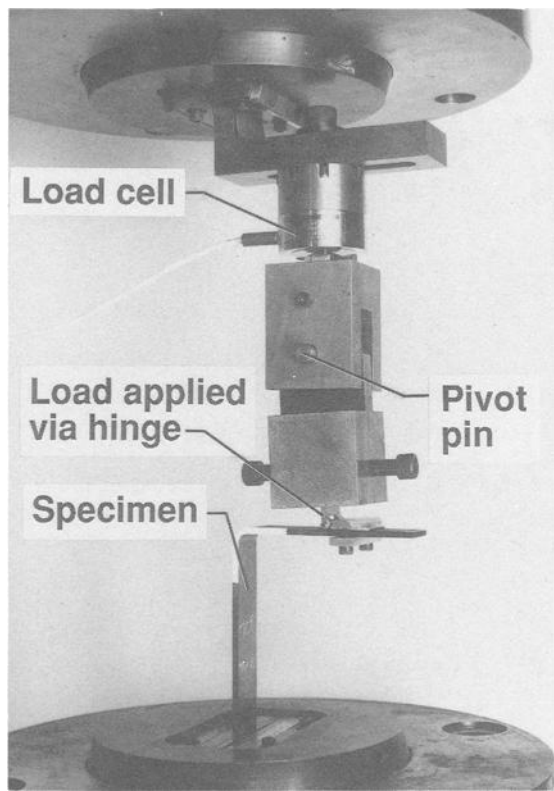


FIG. 3—Curved laminate test fixture.

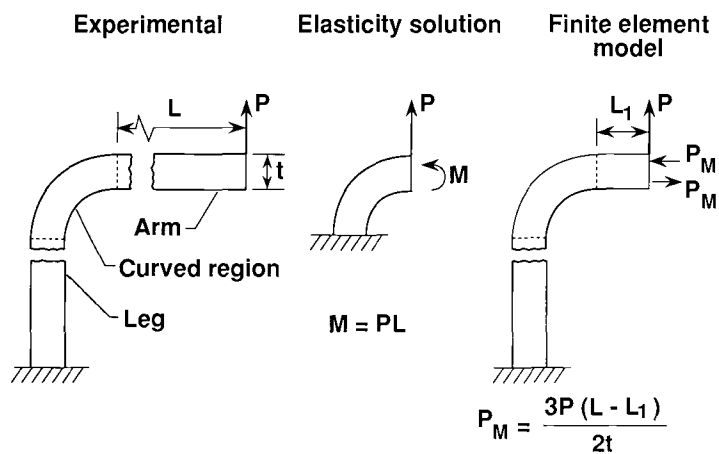


FIG. 4—Curved laminate models.

TABLE 1—Experimental failure loads in the curved laminates.

Panel-Specimen	Maximum Load per Unit Width, N/mm	$\frac{(R - R_{inner})}{t}$ of First Delamination
6-4	7.62	0.49
6-8	14.60	0.48
7-1	8.68	0.46
7-2	10.05	0.48
7-6	13.36	0.48
7-7	13.09	0.35
7-8	7.23	0.33
$7.23 \leq \frac{P_f}{w} \leq 14.60 \text{ N/mm}$		
Mean = 10.66 N/mm		

also given in Table 1. The damage in three different specimens is shown in Fig. 5. Damage initiated as a single delamination around the curved region (see Fig. 5, *top*). Growth was initially unstable and extended into the arm and the leg. Two delaminations around the curved region are shown in Fig. 5 (*center*). The lower delamination is at an $(R - R_{inner})/t$ of 0.23. Final damage was an accumulation of delaminations at different locations in the thickness of the curved region until the specimen lost virtually all bending stiffness. The longest delamination in Fig. 5 (*bottom*) is at an $(R - R_{inner})/t$ of 0.46 and was presumably the first delamination to occur. The upper two photographs shown in Fig. 5 were taken after the load had been reduced upon hearing audible acoustic emissions.

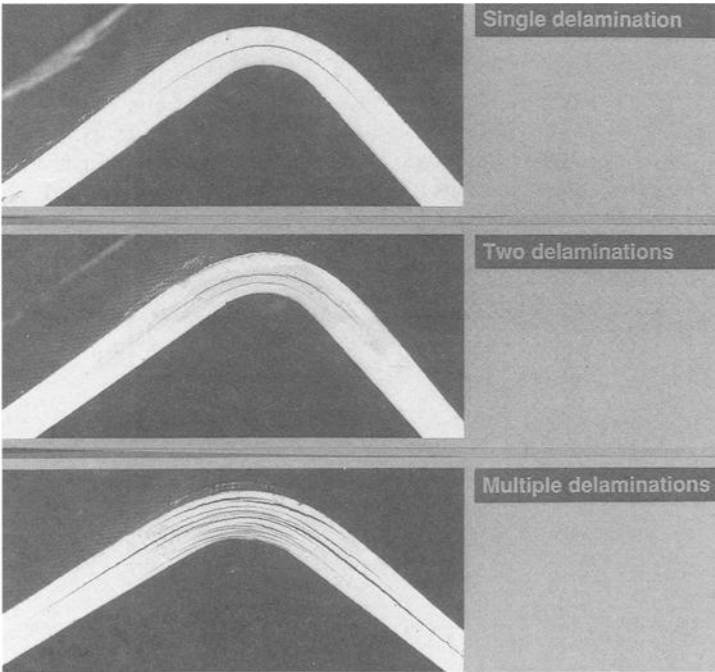


FIG. 3—Damage in undirectional curved laminate.

Analysis

The following section details the analysis required to characterize the onset and growth of a delamination in the unidirectional curved laminate. The location of the highest radial stress was determined using a closed form elasticity solution for an anisotropic curved beam. This analysis was used to determine the location required for mesh refinement in the FEA. A 2-D FEA was conducted to verify the beam solution and determine the variation of G with delamination growth from the location of highest radial stress.

Closed Form Anisotropic Elasticity Solution for a Curved Beam

The loading on the curved laminate used in the experimental work may be reduced to a curved beam with an end load, P , and moment, M , at the free end of a quarter of a curved circular beam (see Fig. 4). The stresses in the beam may be calculated by superimposing the stresses caused by the end load and the moment. A closed form elasticity solution for the stresses in a curved beam with cylindrical anisotropy and subjected to end forces and moments was given in Ref 21 and are shown in Appendix A using the notation given in Fig. 6. The maximum radial stress was determined by solving

$$\frac{d [\sigma_r^P + \sigma_r^M]}{dr} = 0 \tag{1}$$

$$\frac{d [\sigma_r^P + \sigma_r^M]}{d\theta} = 0 \tag{2}$$

for R and θ , where

- σ_r^P = the radial stress created by the end load P and
- σ_r^M = the radial stress created by the moment M .

The solution yielded a maximum value of $[\sigma_r^P + \sigma_r^M]$ at $R = 6.21$ mm and $\theta = 0^\circ$. This value of R was approximated to $R = 6.25$ mm, which was the closest radius to two-ply interfaces. The boundary conditions using this closed form solution are for a quarter circle and not a quarter circle with straight attachments through which the load is transferred. To determine the effect of straight legs attached to a quarter circle, FEA, described in the next section, was conducted.

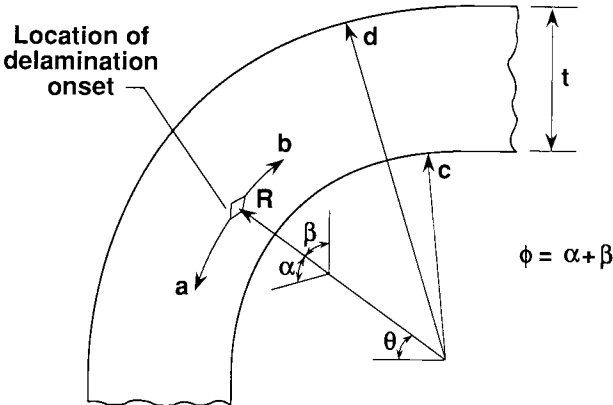


FIG. 6 Notation for unidirectional curved laminate.

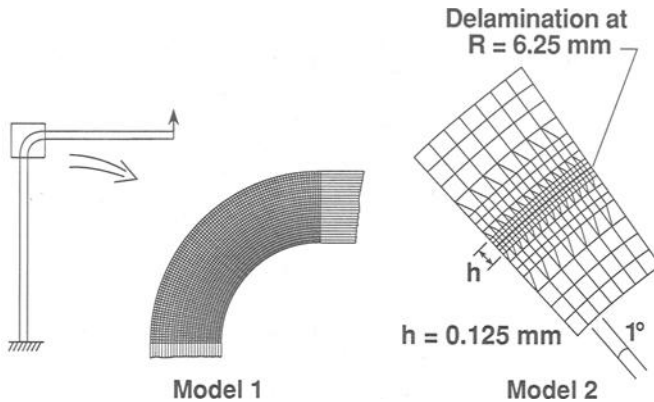


FIG. 7—Finite element mesh.

Finite Element Analysis

A 2-D FEA was conducted using MSC/NASTRAN finite element code [22], two different models, and the dimensions given in Fig. 2. Model 1, shown in Fig. 7 (left), were used to determine the stresses in the curved region of the beam assuming no damage in the part. The model had one element per ply thickness and one element per one degree sweep around the curved region. The model had a unit end load per width (1.0 N/mm) applied at a length L of 50 mm along the arm to simulate the experimental conditions. Model 2 with a refined mesh (Fig. 7, right) was used so that the virtual crack closure technique (VCCT) [23] could be used to determine total G and its components corresponding to delamination growth. The mesh was refined at $R = 6.25$ mm, which was the radius determined from the closed form elasticity solution (and the FEA described later) where delamination was expected to initiate. Near $R = 6.25$ mm, the finite elements measured $h/4$ by $h/4.6$ where h is one ply thickness. At $R = 6.25$ mm, a free surface was included in the model by the use of coincident nodes to represent the delamination. The free surface extended around the complete 90° of the curved region. Connectivity between coincident nodes was established through the use of multipoint constraints (MPCs). By releasing different MPCs in different analysis cases, several delamination lengths could be modeled during one FEA run, yielding an efficient computational technique. To reduce the number of elements in the model, the length of the arm was shortened to $L_1 = 5$ mm. A load and moment were applied to the shortened arm (see Fig. 4). The load and moment were equivalent to an end load of 1 N/mm applied at a moment arm length of 50 mm.

Calculation of Strain Energy Release Rate

The virtual crack closure technique was used to obtain total and individual modal values of G at the delamination fronts. The individual modes of G were calculated at the delamination front using the following equations in polar coordinate notation (Fig. 8)

$$G_1^{VT} = \frac{1}{2\lambda w} F_r^A(\delta_r^B - \delta_r^C) \quad (3)$$

$$G_2^{VT} = \frac{1}{2\lambda w} F_\theta^A(\delta_\theta^B - \delta_\theta^C) \quad (4)$$

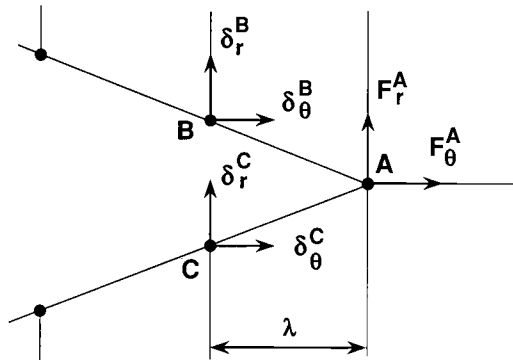


FIG. 8—Virtual crack closure notation.

where

- F_r^A and F_θ^A = the radial and tangential forces at node A, respectively,
- λ = the length of the finite element at the delamination front,
- δ_r and δ_θ = the radial and tangential displacements at the nodes indicated by the superscripts, and
- w = the width of the model.

Total G calculated by VCCT, G^{VT} , was simply a summation of G_I^{VT} and G_{II}^{VT} .

An alternative method of calculating total G to check the VCCT results involved determining the difference in total external work from two finite element runs at two delamination lengths, l_i and l_{i+1} . The total potential is equal to the complementary strain energy, U , which is equal to the internal work done. Therefore, for fixed load conditions:

$$G^{SE} = \frac{\delta U}{\delta a} = \frac{1}{2w(l_i - l_{i+1})} [\Sigma(P\delta)_{l_i} - \Sigma(P\delta)_{l_{i+1}}] \tag{5}$$

where $\Sigma(P\delta)$ is the sum of the product of the applied loads and the displacements (in the loading direction) of the loaded nodes. The resulting G^{SE} approximately applies at a delamination length of $(l_i + l_{i+1})/2$.

Analytical Results

Stress Distribution

Figures 9 and 10 show the radial, transverse shear, and tangential stresses, respectively, as calculated by the closed form elasticity solution and FEA. The stresses were normalized by P/w . The stresses are plotted versus the normalized radius at an angle $\theta = 25^\circ$. The choice of this angle is described below. As shown in Fig. 9, the agreement for the radial and transverse shear stress between the two analytical methods was good and confirms that the radial stress was highest at $(R - R_{\text{inner}})/t = 0.42$ ($R \approx 6.25$ mm). The transverse shear stress was approximately an order of magnitude smaller than the radial stress. The agreement in Fig. 10 was also good and shows that the laminate has a maximum tensile tangential stress at the inside edge of the curved region.

Figure 11 shows the normalized radial stress plotted around the radius at $R = 6.25$ mm. The closed form elasticity solution agreed well with the FEA for $20^\circ \leq \theta \leq 60^\circ$. However, as

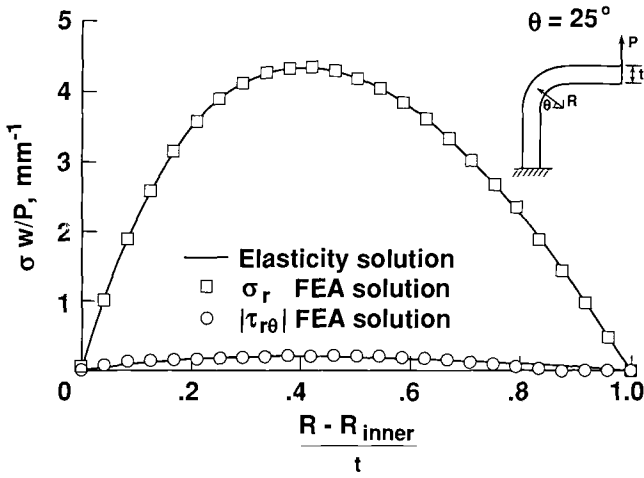


FIG. 9—Interlaminar stresses through thickness.

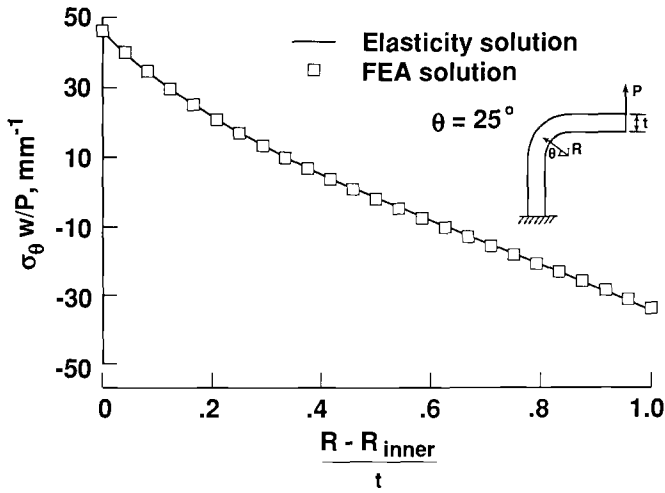


FIG. 10—Tangential stress through thickness.

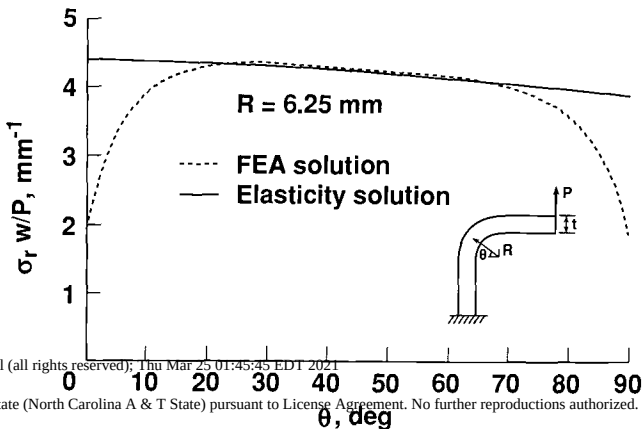


FIG. 11—Radial stress around curved region.

θ approaches the boundaries, the two solutions diverge, as expected from St. Venant's principle. According to the FEA, the radial stress reaches a maximum at $\theta = 25^\circ$. Therefore, the location for the onset of delamination was determined from Figs. 9 through 11 to be at $R = 6.25$ mm at an angle $\theta = 25^\circ$.

Strain Energy Release Rate Variation with Delamination Growth

Next, the variation of G with delamination growth was determined, assuming the delamination to have occurred at the location of highest radial stress. The delamination had two fronts, which may have different values of G . Referring to Fig. 6, the delamination front may grow counter-clockwise or clockwise from the initiation point ($R = 6.25$ mm, $\theta = 25^\circ$) as indicated by the letter a or b , respectively. The delamination length is given by $a + b$ where

$$a = \frac{2\pi R\alpha}{360} \quad \text{and} \quad b = \frac{2\pi R\beta}{360}$$

and α and β are measured in degrees.

Figure 12 shows the normalized G distribution (Gw^2/P^2) with delamination growth in the α direction only, while holding β equal to zero. The delamination was predominantly Mode I around the curved region. The normalized values of G_I^{VT} obtained by VCCT ($G_I^{VT} + G_{II}^{VT}$) closely agree with values obtained using the change in global strain energy technique, G^{SE} . Both G^{SE} and G_I^{VT} increase to a maximum at $\alpha = 22^\circ$ and then reduce. At $\alpha = 25^\circ$, the delamination is at the curved region/leg boundary.

Figure 13 shows the normalized G distribution in the β direction, holding α equal to zero. Again, the delamination growth was predominantly Mode I around the curved region. The total G results from VCCT and the change in global strain energy technique again agree well. Both G_I^{VT} and G^{SE} increase from zero to a maximum and then decrease. For G_I^{VT} , the maximum is at $\beta = 48^\circ$ while for G^{SE} the maximum is at $\beta = 50^\circ$. The difference between the two peaks is caused by the increase in G_{II}^{VT} .

Figures 12 and 13 show how G varies when either one of the two fronts grows individually. However, they do not show how the two fronts grow simultaneously. Consider an angle ϕ , where $\phi = \alpha + \beta$. Also, assume that the delamination grows equally in the a and b direction,

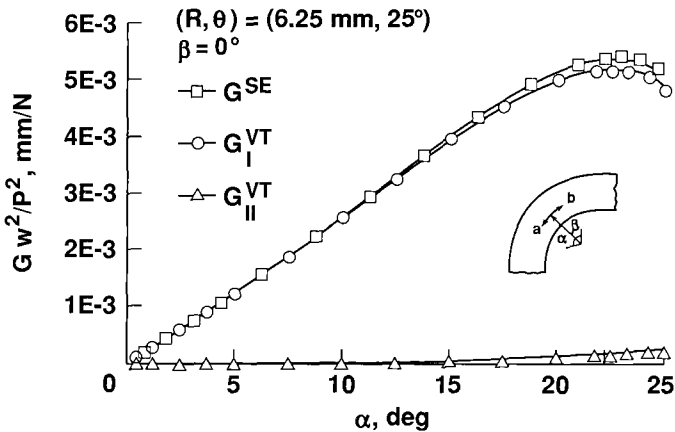
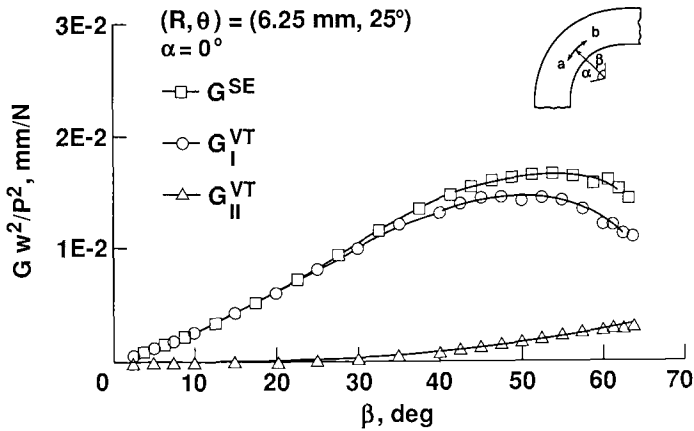
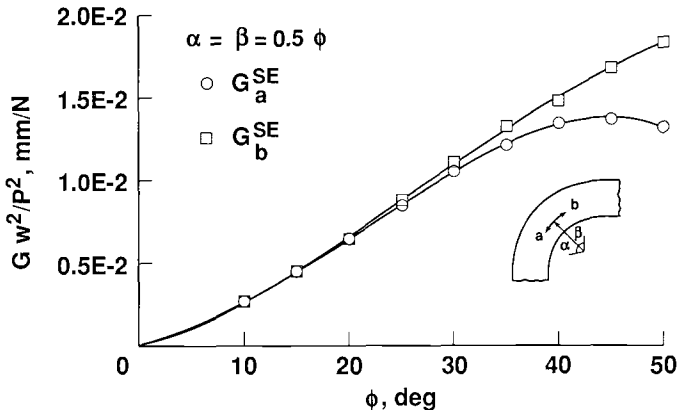


FIG. 12 G variation with delamination growth in the α direction.

FIG. 13— G variation with delamination growth in the β direction.

hence, $\alpha = \beta = \phi/2$. The result of assuming this delamination growth behavior is given in Fig. 14. Note that when $\phi = 50^\circ$, the delamination has grown fully around the curved region to the leg but has not yet reached the arm (see Fig. 6). For small angles of ϕ , G_a and G_b are similar and the delamination should grow equally in the α and β direction. However, at larger angles of ϕ , G_a becomes less than G_b because σ_r is decreasing more rapidly in the α direction than in the β direction (see Fig. 11). Hence, the delamination would continue to grow in the β direction and not in the α direction. Thus, any subsequent data points in Fig. 14 are invalid. Delamination growth must be represented incrementally, and G_a and G_b must be compared at each increment. If G at one end of the delamination is higher than that at the other, then the delamination will grow in that direction only. After that incremental growth, another analysis must be conducted, and so on until the delamination arrests or failure occurs.

Hence, a method was developed to ensure a constant G existed at both delamination fronts. A criterion was used that if G was 2% greater at one delamination front than at the other, then growth would occur in that direction only. If the difference was less than 2%, then the delamination would grow in both directions simultaneously. For the first increment, a 5° step was

FIG. 14— G variation with delamination growth in the α and β directions.

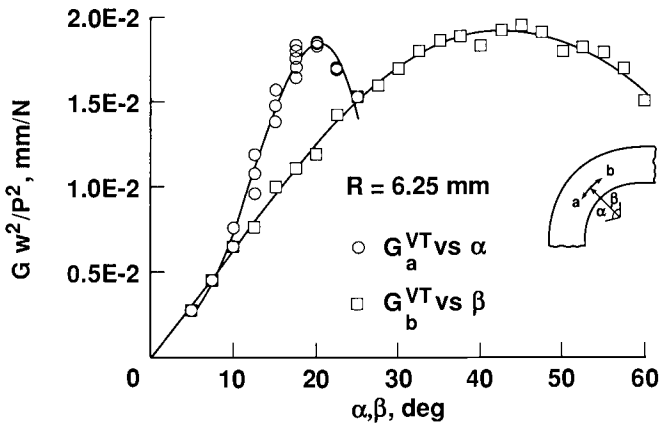


FIG. 15— G_a versus α and G_b versus β .

taken from the initiation site. From there, incremental delamination growth steps of 2.5° or 0.273 mm were assumed. The results are shown in Fig. 15. Note that G_a is plotted against α , which ranges from 0 to 25° , whereas G_b is plotted against β , which ranges from 0 to 65° . From the onset location, the delamination grows simultaneously in both directions as indicated in Fig. 14. However, at the third increment, G_b was 3.7% higher than G_a . Hence, delamination growth was assumed to occur in the β direction alone and is shown in Fig. 15 by the next G_a value being plotted directly above the previous one at the same value of α . At this increment, G_a and G_b have less than a 2% difference. Hence, delamination growth occurs in both directions simultaneously shown by the next G_a and G_b values being plotted at an increased value of α and β , respectively. This procedure was continued until the delamination reached the leg.

A plot of normalized G^{SE} versus ϕ is shown in Fig. 16 using the results of Fig. 15. This plot details the variation in normalized G , with delamination growth taking into account the two delamination fronts. The delamination grows unstably until it has grown 60° around the curve. Total G then decreases and the growth becomes stable. Figure 17 shows the variation of the G_I/G ratio as the delamination grows. Delamination growth is predominantly Mode I, with pure Mode I delamination growth at small delamination lengths.

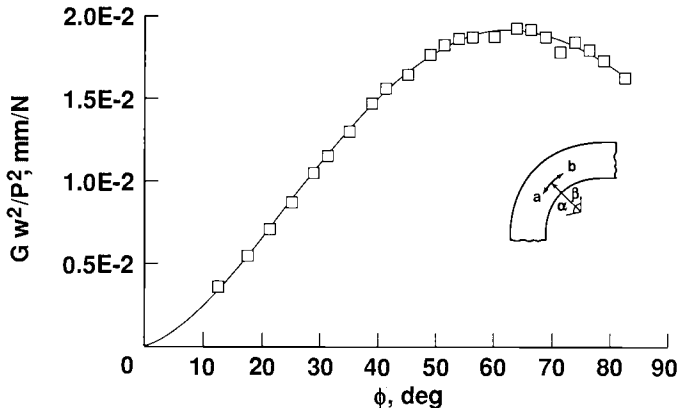
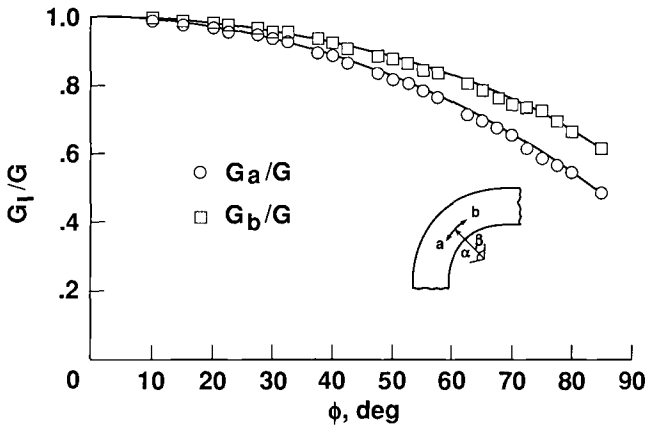
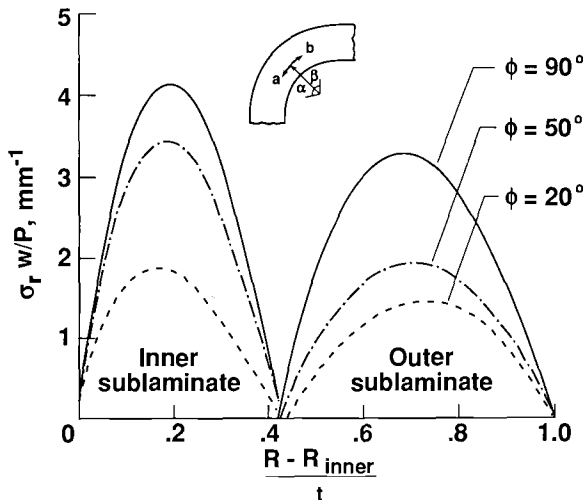


FIG. 16— G variation with delamination growth in both directions.

FIG. 17— G_I/G ratio with delamination growth.

Damage Progression in the Unidirectional Curved Laminates

Figure 18 shows the normalized radial stresses plotted versus the normalized radius of the curved laminate with a delamination present at $(R - R_{\text{inner}})/t = 0.42$. Three specific delamination lengths taken from Fig. 16 are shown. As the delamination grows around the curved region, the maximum radial stress in the newly forming inner and outer sublaminates increases. At $\phi = 90^\circ$, G is decreasing (see Fig. 16) and the maximum radial stress in the inner sublaminate has reached a value similar to that responsible for the formation of the original delamination at $R = 6.25$ mm as shown in Fig. 9. Therefore, a new delamination should start in the inner sublaminate as was observed experimentally (see Fig. 5). No further analysis of delamination growth from this location was conducted, but it was assumed that this phenomenon would repeat sequentially and result in the final failure shown in Fig. 5.



Failure Load Predictions and Discussion

To determine the maximum loads required to initiate damage in the unidirectional curved laminates, a suitable failure criterion must be used. Because of the high tensile strength in the fiber direction it is not necessary to consider the σ_θ stresses. Also, the values of $\tau_{\theta\theta}$ are an order of magnitude smaller than the σ_r stresses and may be considered negligible. Therefore, it is sufficient for a strength approximation to assume that delamination occurs when $\sigma_r = \sigma_{3f}$, where σ_{3f} is the interlaminar normal strength. However, as noted in the Introduction, no standard test for determining interlaminar normal strength has been developed. Hence, transverse isotropy must be assumed and σ_{3f} must be approximated by σ_{2f} . From Fig. 9, the maximum normalized radial stress was given as $(\sigma_r w/P) = 4.37 \text{ mm}^{-1}$ occurring at an $(R - R_{\text{inner}})/t = 0.42$. Therefore, a range of damage initiation loads per unit width can be determined

$$\frac{P_f}{w} = \frac{\sigma_{2f}}{(\sigma_r w/P)} = 8.36 \leq \frac{P_f}{w} \leq 15.8 \text{ N/mm} \quad (\text{mean} = 12.3 \text{ N/mm})$$

Table 1 shows that the agreement between the range of loads and the range of predicted loads was good. The mean failure load was overpredicted by 15%. The predicted location of the delamination through the thickness was similar to that observed in the experimental work (see Table 1).

The plot of normalized G versus ϕ shown in Fig. 16 shows G beginning to decrease at $\phi \approx 60^\circ$. This type of decrease in G with delamination growth is often interpreted as an indication of stable delamination growth. However, the delamination usually grew well into the arm and leg (see Fig. 5). Delamination growth was shown in Fig. 17 to be largely Mode I dominated. Therefore, using the criterion that $G = G_c = G_{lc}$ at failure, a predicted load per unit width to further extend the delamination may be calculated. From Fig. 16 at $\phi = 90^\circ$, $(Gw^2/P^2) = 1.60$. $E - 2$ or 0.0160 mm/N .

$$\frac{P_f}{w} = \left[\frac{G_{lc}}{\left(\frac{Gw^2}{P^2} \right)} \right]^{1/2} = 2.01 \leq \frac{P_f}{w} \leq 2.41 \text{ N/mm} \quad (\text{mean} = 2.24 \text{ N/mm})$$

where G_{lc} was obtained from the DCB specimens. These values of applied load per unit width are well below the load already applied to the curved laminate, indicating that the delamination will extend into the arm and the leg until G decreases below G_c .

At a delamination length of $\phi = 90^\circ$, the maximum normalized radial stress in the inner sublaminates from Fig. 18 was $(\sigma_r w/P) = 4.10 \text{ N/mm}^{-1}$ at an $(R - R_{\text{inner}})/t \approx 0.20$. A new delamination occurs at a load per unit width between 8.90 and 16.8 N/mm. This range of loads is approximately equal to the range of loads already on the specimen. Also, the predicted location of the second delamination is similar to that shown in Fig. 5 (center). Hence, damage accumulating at different locations through the thickness would be expected to be unstable, as observed experimentally.

Summary

This paper investigated delamination failure in curved composite laminates. A unidirectional curved laminate was tested quasistatically and analyzed. These laminates failed unstably by delaminations radiating around the curved region of the laminate and accumulating at different locations through the thickness until most of the bending stiffness was lost. Delami-

nation was assumed to initiate at the location of the highest radial stress within the curved region. A closed form curved beam elasticity solution and FEA were used to determine the location of the highest radial stress. A strength-based failure criterion predicted the failure load for the onset of the first delamination. The FEA was then used to determine the change in strain energy release rate, G , with delamination growth in both directions from the initiation site. The strain energy release rate was found to be predominantly Mode I. Strain energy release rate initially increased with delamination growth and then decreased as the delamination reached the straight arm and leg. However, the value of G with delamination fully extended around the curved region was larger than the critical value of G determined from double cantilever beam specimens. Thus, delamination growth was predicted to extend into the arm and leg of the laminate as observed experimentally. As the initial delamination grew around the curved region, the maximum radial stress in the newly forming inner sublaminar increased to a sufficient level to cause a new delamination to initiate in the sublaminar with no increase in applied load. This failure sequence was also observed experimentally.

APPENDIX A

Closed Form Elasticity Solution for the Stresses in an Anisotropic Curved Beam

Reference 22 gave the elasticity solutions for the radial, tangential, and shear stresses in a curved beam subjected to pure bending and an end force. The beam was assumed to possess cylindrical anisotropy and its pole is located at the center of the arcs that form the beam contour. The expressions are repeated here with the notation given in the Nomenclature section and in Fig. 6.

Stresses Caused by End Moment M

$$\sigma_r^M = \frac{-M}{d^{2wg}} \left[1 - \frac{1 - \left(\frac{c}{d}\right)^{\kappa+1}}{1 - \left(\frac{c}{d}\right)^{2\kappa}} \left(\frac{r}{d}\right)^{\kappa-1} - \frac{1 - \left(\frac{c}{d}\right)^{\kappa-1}}{1 - \left(\frac{c}{d}\right)^{2\kappa}} \left(\frac{c}{d}\right)^{\kappa+1} \left(\frac{d}{r}\right)^{\kappa+1} \right] \quad (\text{A1})$$

$$\sigma_\theta^M = \frac{-M}{d^{2wg}} \left[1 - \frac{1 - \left(\frac{c}{d}\right)^{\kappa+1}}{1 - \left(\frac{c}{d}\right)^{2\kappa}} \kappa \left(\frac{r}{d}\right)^{\kappa-1} + \frac{1 - \left(\frac{c}{d}\right)^{\kappa-1}}{1 - \left(\frac{c}{d}\right)^{2\kappa}} \kappa \left(\frac{c}{d}\right)^{\kappa+1} \left(\frac{d}{r}\right)^{\kappa+1} \right] \quad (\text{A2})$$

$$\tau_{r\theta}^M = 0 \quad (\text{A3})$$

where

$$g = \frac{1 - \left(\frac{c}{d}\right)^2}{2} - \frac{\kappa}{\kappa + 1} \frac{\left[1 - \left(\frac{c}{d}\right)^{\kappa+1}\right]^2}{\left[1 - \left(\frac{c}{d}\right)^{2\kappa}\right]} + \frac{\kappa \left(\frac{c}{d}\right)^2 \left[1 - \left(\frac{c}{d}\right)^{\kappa-1}\right]^2}{\kappa - 1 \left[1 - \left(\frac{c}{d}\right)^{2\kappa}\right]} \quad (\text{A4})$$

and the anisotropic parameter is defined as

$$\kappa = \left[\frac{E_\theta}{E_r} \right]^{1/2} \quad (\text{A5})$$

Stresses Caused by End Force P

$$\sigma_r^P = \frac{P}{rwg_1} \left[\left(\frac{r}{d} \right)^\omega + \left(\frac{c}{r} \right)^\omega - 1 - \left(\frac{c}{d} \right)^\omega \right] \cos \theta \quad (\text{A6})$$

$$\sigma_\theta^P = \frac{P}{rwg_1} \left[(1 + \omega) \left(\frac{r}{d} \right)^\omega + (1 - \omega) \left(\frac{c}{r} \right)^\omega - 1 - \left(\frac{c}{d} \right)^\omega \right] \cos \theta \quad (\text{A7})$$

$$\tau_{r\theta}^P = \frac{-P}{rwg_1} \left[\left(\frac{r}{d} \right)^\omega + \left(\frac{c}{r} \right)^\omega - 1 - \left(\frac{c}{d} \right)^\omega \right] \sin \theta \quad (\text{A8})$$

where

$$g_1 = \frac{2}{\omega} \left[1 - \left(\frac{c}{d} \right)^\omega \right] + \left[1 + \left(\frac{c}{d} \right)^\omega \right] \ln \left(\frac{c}{d} \right) \quad (\text{A9})$$

and the anisotropic parameter, ω , is defined as

$$\omega = \left[1 + \frac{E_\theta}{E_r} (1 - 2\nu_{r\theta}) + \frac{E_\theta}{G_{r\theta}} \right]^{1/2} \quad (\text{A10})$$

Acknowledgments

This work was supported by NASA Contract NAS1-18599

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Efficient Use of Film Adhesive Interlayers to Suppress Delamination

REFERENCE: Lagace, P. A. and Bhat, N. V., "Efficient Use of Film Adhesive Interlayers to Suppress Delamination," *Composite Materials: Testing and Design (Tenth Volume)*, ASTM STP 1120, Glenn C. Grimes, Ed., American Society for Testing and Materials, Philadelphia, 1992, pp. 384–396.

ABSTRACT. An experimental and analytical study was conducted to assess the effectiveness of the efficient use of film adhesive interlayers in suppressing delamination in composite laminates. Efficiency was achieved by two means: (1) the adhesive interlayer was placed only at delamination critical interfaces, identified by calculating the interlaminar stresses and applying the Quadratic Delamination Criterion, in comparison to all ply interfaces as in previous investigations; and (2) finite-width strips were used rather than full-width film adhesive layers. Graphite/epoxy [$\pm 15^\circ/0^\circ$]_s laminates, with FM300 Interleaf material placed at the $+15^\circ/-15^\circ$ interface, were tested in uniaxial tension. For all specimens with film adhesive interlayers, the load at which delaminations initiated was about 50% higher than in the case of the specimens without film adhesive. The final failure stresses of the specimens with film adhesive also averaged 40% higher than those without the interlayers. These results were not discernably affected by the use of a reduced number of interlayers and, furthermore, did not depend upon the width of the strip. This indicates that, within practical manufacturing limitations, any strip which is as large as the interlaminar stress boundary layer would be successful in inhibiting delamination initiation. However, in tests which Teflon® strips were inserted to simulate preinitiated delamination, the results were similar to those in which no film adhesive interlayers were present. This indicates that the interlayers are not successful in suppressing the growth of delamination which is already present. Experimental results also suggest the possible existence of a "critical" delamination initiation size.

KEYWORDS. delamination, interlaminar stress, fracture, damage suppression, graphite/epoxy

Nomenclature

σ_{1z}, σ_{2z}	Interlaminar shear stresses
σ_{zz}	Interlaminar normal stress
σ_{11}	Applied loading
x	Distance from the free edge
t_{ply}	Nominal ply thickness
FA	FM300 interleaf film adhesive material

Introduction

An important failure mode in composite laminates is delamination. Delamination refers to the failure of the interply matrix layer and is caused by interlaminar stresses [1] which arise as

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a result of the compliance mismatch between plies at free edges [2] or in any gradient stress field [3] The occurrence of delamination can increase significantly the structural compliance of a component and lead to the ultimate failure of the part via interaction with other failure modes This failure via delamination often occurs at loads well below those achievable if only in-plane damage and failure mechanisms are at work [4,5]

A number of approaches have been used to prevent and limit delamination (for example, Refs 6 and 7) However, the most effective of these appears to address the initiation of delamination If delamination is not allowed to initiate, then the resultant increase in compliance and reduction in load-carrying capability of the structure cannot occur The problem of suppressing or delaying delamination initiation has been addressed in two ways The first approach aimed at strengthening the interface between plies of the laminate (for example, Refs 6 and 7), and the second aimed at reducing the interlaminar stresses in the laminate (for example, Refs 8 and 9) Since the greatest success has been achieved with the latter technique, this is the subject of the current paper

A compliant layer such as film adhesive co-cured between the plies of the laminate has been shown to be successful in preventing delamination initiation [8,9] These interlayers mitigate the compliance mismatch between the plies and, hence, reduce the high interlaminar stresses that arise, thus delaying or totally preventing delamination Earlier work on this topic did not consider the efficiency of film adhesive placement in the effort to suppress delamination Lagace et al [9] placed film adhesive at all dissimilar interfaces of the laminate Chan et al [8] showed that strips of film adhesive at the free edge could suppress delamination, but the strips of film adhesive were of arbitrary size

The primary objective of the current study is to examine the efficient use of film adhesive in the suppression of delamination Efficiency is examined in two ways one, by reducing the number of interfaces at which the interlayer material is placed and two, by using finite width strips of the interlayer material in the regions with high interlaminar stresses

Scope

The two methods by which efficiency is achieved are determined by considering the interlaminar stress state. For the first method, the film adhesive strips are placed only at "delamination critical" interfaces, which are defined as the interfaces where delamination is most likely to initiate This is determined by calculating the interlaminar stress state and then using a delamination initiation criterion to identify the interface(s) where delamination is likely to occur In the present work, the interlaminar stress state was calculated via the method of Kasapoglou and Lagace [10] as modified by Saeger and Lagace [11] (the Reduced Eigenfunction Technique), and then the Quadratic Delamination Criterion [12] was applied to determine the critical interfaces

The second efficiency method involved considering the region near the free edge where the interlaminar stresses dominate, which is referred to as the boundary layer region [10]. Since the interlaminar stresses are confined to a small boundary layer region near a free edge in a laminate, it is reasonable to assume that the film adhesive is needed only in this region, and thus, the material is restricted to the free edge region in the form of strips Strip sizes were chosen as fractions of the boundary layer size to examine the effect of strip width on the delamination behavior of the specimens The results from these specimens were to be compared to specimens with whole plies of film adhesive at the delamination critical interface The use of the film adhesive as strips at the free edge requires that these strips be "dropped off" or terminated within the laminate This creates another location at which interlaminar stresses would arise The termination of the strip of film adhesive within the free edge boundary layer region of the laminate could lead to interference of the two types of interlaminar stress fields,

one as a result of the free edge and the other a result of the drop-off. The use of different strip widths should provide an insight into the significance of the interference of these stresses and whether it does affect the delamination behavior of the specimen.

The experimental program consists of testing $[\pm 15_n/0_n]_s$ laminates made from AS4/3501-6 graphite/epoxy. This particular lay-up was chosen because it has earlier been shown to exhibit significant delamination behavior [13]. The $\pm 15^\circ/-15^\circ$ interface has been shown to delaminate, leading to premature failure at loads much lower than the in-plane failure load predicted by a typical in-plane criterion (for example, Ref. 14). The Quadratic Delamination Criterion [12] also predicts that the delamination will initiate at the $+15^\circ/-15^\circ$ interface, hence, this interface is termed as the delamination critical interface of the laminate.

An analytical study using the Reduced Eigenfunction Solution Technique [11] and the Quadratic Delamination Criterion [12] was performed on the $[\pm 15_n/0_n]_s$ laminate to compare the delamination initiation behavior of the laminate when the film adhesive was placed at all dissimilar interfaces of the laminate to the case in which the film adhesive was placed only at the delamination critical interface ($+15^\circ/-15^\circ$). The results shown in Table 1 indicate that the stresses at which delamination is predicted to initiate for both the cases are within 1% of each other. The stresses presented in Table 1 are normalized with respect to the cross-sectional area of the laminate without any film adhesive for purposes of comparison.

The interlaminar shear stress σ_{1z} is the primary cause of free edge delamination in this case as it is an order of magnitude larger than either the σ_{zz} or σ_{2z} stresses [12, 15]. The presence of the film adhesive at the $+15^\circ/-15^\circ$ interface results in a significant reduction in the magnitude of this stress in the vicinity of the free edge as shown in Fig. 1. In the Quadratic Delamination Criterion, a stress within a characteristic dimension (0.178 mm for AS4/3501-6 graphite/epoxy) of the free edge is averaged and used. This leads to an increase in the predicted delamination initiation load as a result of the reduction in interlaminar stress magnitude in this region. Hence, the first step towards efficient usage of film adhesive is to place the film adhesive only at the delamination critical interface.

To evaluate effectively the second efficiency scheme, a laminate with a sufficiently large boundary layer needed to be considered. It has previously been shown that the extent of the boundary layer region increases linearly with effective ply thickness [13]. A normalized (via the ply thickness) effective ply thickness (value of n) of seven was chosen to increase the boundary layer size to about 9 mm, thus making it possible to manufacture film adhesive strips with widths that were fractions of the boundary layer size. This results in a $[\pm 15_7/0_7]_s$ laminate.

Five different types of specimens were considered in evaluating this efficiency technique. The first type were control specimens, $[\pm 15_7/0_7]_s$, with no film adhesive in them which were used to obtain the basic laminate behavior specifically in terms of delamination initiation and final failure. The next four specimen types had film adhesive strips at the $+15^\circ/-15^\circ$ interface at the free edge of the laminate. The coupon used for the test is shown in Fig. 2. The strip widths used were 25, 9, 6, and 3 mm. The test coupon is 50 mm wide, thus, a 25-mm strip at each

TABLE 1—Calculated effect of film adhesive placement

Laminate	Delamination Initiation Stress, MPa	In-Plane Fracture Stress, MPa
$[\pm 15/-15/0]_s$	819	1790
$[\pm 15/FA^a/-15/FA/0]_s$	1377	1911
$[\pm 15/FA/-15/0]_s$	1349	1857

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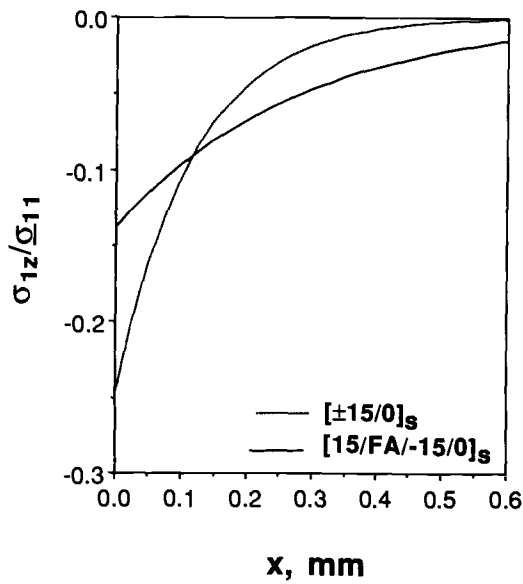


FIG. 1—Distribution of interlaminar shear stress, σ_{12} , at the $+15^\circ/-15^\circ$ interface in a $[\pm 15/0]_s$ laminate and the $15^\circ/FA$ interface in a $[15/FA/-15/0]_s$ laminate.

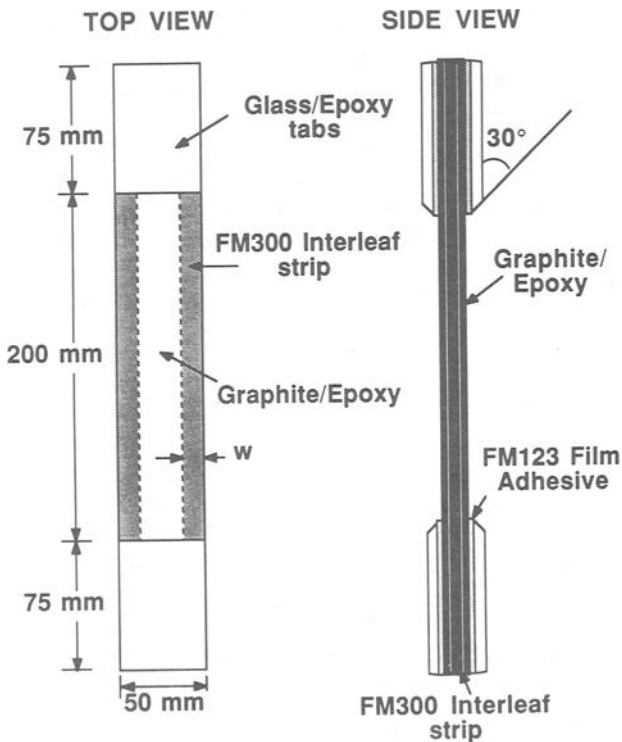


FIG. 2—Physical characteristics of the coupon specimen.

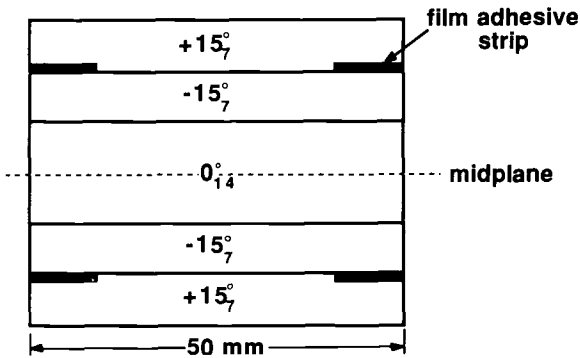


FIG. 3—Illustration of cross section of coupon with film adhesive strips.

free edge implies a complete ply of film adhesive through the width of the specimen. A cross-sectional illustration of the test specimen with the film adhesive strips is shown in Fig. 3.

An additional objective of this investigation is to isolate the effect of the film adhesive on delamination growth versus delamination initiation. To investigate delamination growth experimentally, specimens were made with implanted delaminations. These specimens had full-width film adhesive plies at the +15°/−15° interface of the laminate, and strips of Teflon,[®] 0.08 mm thick, were placed at the free edge of the +15°/FA interface to simulate preinitiated delamination. Since the Teflon strip represents an initiated delamination, the failure stress of the specimen would indicate the effect of the film adhesive on the growth of the delamination alone. The Teflon strips were chosen to be of various sizes, in relation to the boundary layer size, to examine the effect of delamination initiations of different size. The different strip widths were 9, 6, and 3 mm. Their placement in the specimen cross section is illustrated in Fig. 4.

The complete testing program is shown in Table 2. All laminates were made from Hercules AS4/3501-6 graphite/epoxy with a nominal ply thickness of 0.134 mm. The film adhesive was American Cyanamid FM300 Interleaf, chosen for its co-cure compatibility with the 3501-6 system and also its minimal flow during the cure cycle. This material has a nominal thickness, after cure, of 0.203 mm. The material properties of these materials are given in Table 3.

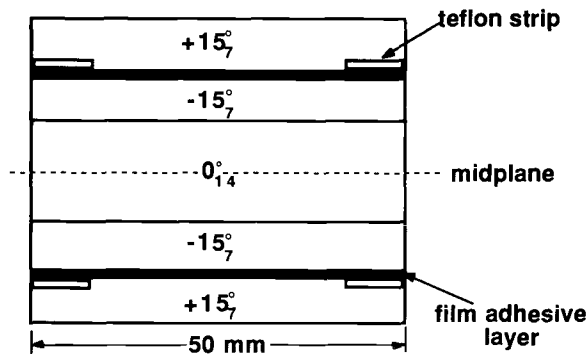


FIG. 4—Illustration of cross section of coupon with film adhesive ply and Teflon implants.

TABLE 2—*Testing program*

Width of Film Adhesive, mm	Width of Teflon Strip, mm	Number of Specimens
25 ^a		4
9		4
6		4
3		4
25 ^a	9	4
25 ^a	6	4
25 ^a	3	4

^a Indicates full ply of film adhesive

Experimental Procedure

Each coupon was manufactured individually, as considerable movement and, hence, misalignment of the film adhesive strips occurred when using laminates which were subsequently cut into four specimens. A slight variation of the standard AS4/3501-6 cure cycle was used. It consists of a 1-h flow stage at 116°C and a 2-h set stage at 177°C, all under vacuum. A 0.59-MPa external pressure was also applied 40 min into the flow stage. This variation in the time when the pressure was applied was used to minimize the movement of the film adhesive strips during the flow stage of the cure cycle. The laminates were postcured at 177°C for 8 h.

All coupons were cut to their final width of 50 mm using a water-cooled diamond wheel. The free edges of the specimens, except those with Teflon inserts, were carefully polished to increase the clarity of the edge replicas taken to detect the initiation of delamination. Polishing was done with a 25-mm felt bob mounted in a drill press. The felt bob was continually dipped in a colloidal solution of a fine abrasive, Kaopolite SF, which was an average particle size of 0.7 μm . A smooth back-and-forth motion was used while the free edge was passed against the felt bob. In addition, each of the 32 specimens had a longitudinal and transverse strain gage.

The presence of the film adhesive at the edges of the specimens caused a 4% increase in the thickness at the edges compared to a 7% increase expected from nominal thickness calculations. The film adhesive layer, however, showed considerable variation in thickness along the length of the specimen. The thickness of the film adhesive varied from virtually 0 to 180% of the nominal thickness of 0.203 mm. This variation of the film adhesive is indicated in the optical micrograph of the edge shown in Fig. 5.

TABLE 3—*Basic material properties of AS4/3501-6 and FM300 interleaf*

Property	AS4/3501-6 Graphite/Epoxy	FM300 Interleaf Film Adhesive
E_{11}	142.0 GPa	5.17 GPa
E_{22}	9.8 GPa	5.17 GPa
E_{zz}	9.8 GPa	5.17 GPa
G_{12}	6.0 GPa	1.83 GPa
G_{1z}	6.0 GPa	1.83 GPa
G_{2z}	4.8 GPa	1.83 GPa
ν_{12}	0.30	0.388
ν_{1z}	0.30	0.388
ν_{2z}	0.57	0.388
t_{ply}	0.134 mm	0.203 mm ^a

^a Made up via film adhesive plies of three different thicknesses (0.041, 0.051, and 0.102 mm) as supplied by the manufacturer.

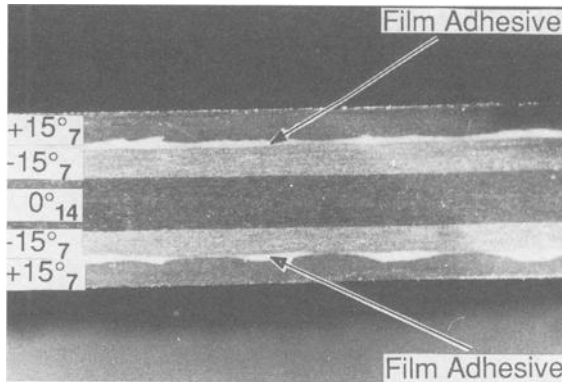


FIG. 5—Photomicrograph showing the variation of thickness of the implanted film adhesive along the edge of the specimen.

All specimens were tested under computer control in a Material Test System MTS 810 testing machine equipped with hydraulic grips. A stroke rate of 1.1 mm/min yielded a strain rate of approximately 5000 microstrain/minute in the test section. Load, strain, and stroke data were recorded automatically through a data acquisition system.

In the specimens tested for delamination initiation, the load drop technique was used. The testing machine was immediately stopped under computer control when it was determined that the applied load had dropped from the previous data point. This load drop can be indicative of the loss in modulus as a result of delamination [16]. The sensitivity to such load drops is dependent on the size of the initiation, the reduction in modulus in the delaminated region, the frequency of the data acquisition, the resolution in the load measurement, and the noise in the data acquisition system. The data acquisition rate was selected to minimize the chances of missing the load drops which accompany delamination initiation while also reducing the number of tests inadvertently stopped by noise in the data acquisition system [17].

Delamination initiation was confirmed using an edge replication technique. After each test, the stroke was reduced by a factor of two, and an edge replica was made. Acetate tape was softened with acetone and then smoothed against the specimen edge. When examined under a microscope with backlighting at magnifications of $\times 5$ to $\times 20$, delamination initiation could be easily detected. These replicas were compared to replicas from the previous test or ones made before any tests were run. If no initiation was found, the load drop was assumed to have been caused by noise in the data acquisition system, and the specimen was retested. If initiation was found, the stress corresponding to the maximum load was taken to be the initiation stress. The edge replication technique is described in detail in Ref 17. In such laminates, delamination initiation is observed to occur at the $+15^\circ/-15^\circ$ interface before any transverse cracking in the adjacent plies [12].

Since the elastic modulus of the FM300 Interleaf material (5.17 GPa) is small compared to the longitudinal modulus of the graphite/epoxy (142 GPa), the interleaf material carries a very small portion of the total in-plane longitudinal load in these laminates. Hence, only the nominal thickness of the graphite/epoxy was used in calculating the cross-sectional area of the specimen for converting load data to stress. This allowed direct comparison of the stresses between the various types of specimens involved.

Results and Discussions

The experimental results are summarized in Table 4 in which the average delamination initiation loads and the average fracture stresses of the specimens are shown.

TABLE 4—Summary of experimental results.

Width of Film Adhesive, mm	Width of Teflon Strips, mm	Delamination Initiation Stress, MPa (%)	Fracture Stress, MPa (%)
...	...	411 (10.4) ^a	486 (3.9)
25 ^b	...	646 (5.2)	679 (8.4)
9	...	647 (10.1)	718 (4.1)
6	...	605 (19.9)	694 (5.7)
3	...	667 (12.8)	708 (5.2)
25 ^b	9	^c	459 (8.0)
25 ^b	6	^c	450 (6.3)
25 ^b	3	^c	591 (6.1)

^a Numbers in parentheses are coefficients of variation.

^b Indicates full ply of film adhesive.

^c Does not apply; delamination is preinitiated.

Delamination Initiation

The control specimens exhibited delamination initiation at the +15°/−15° interface at a stress of 411 MPa. This compares well with the predicted delamination initiation stress of 466 MPa using the Quadratic Delamination Criterion. The failed specimens show extensive delamination and relatively little fiber failure (in that many of the 0° fibers were not broken), as can be seen in Fig. 6. The in-plane failure load predicted using the Stress Interaction Criterion is 1790 MPa, which is much higher than that achieved by the specimen. This can be attributed to the out-of-plane damage exhibited by the specimens.

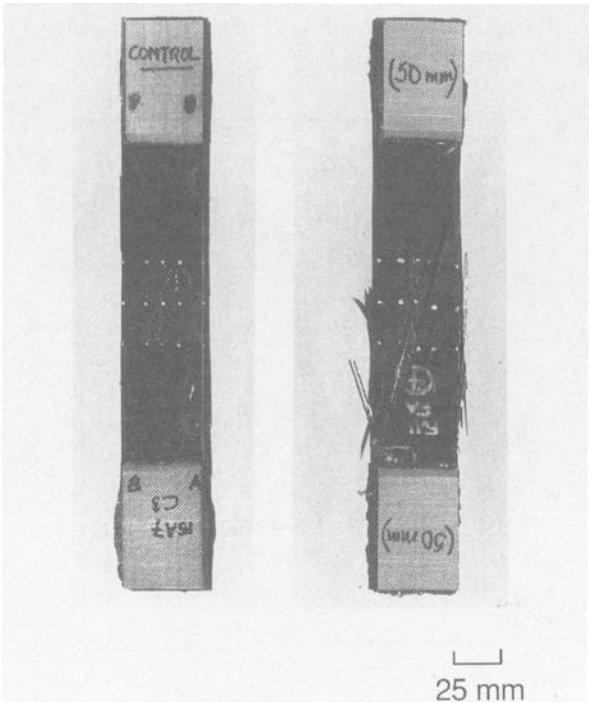


FIG. 6—Photograph of typical failures of (left) a $[\pm 15/0]_s$ control specimen and (right) a $[15/F/A/15/0]_s$ specimen.

The specimens with a full-width ply of film adhesive exhibited delamination initiation at an average stress of 646 MPa, which is about 50% higher than the delamination initiation stress of the control specimens. The predicted delamination initiation stress for these specimens is 639 MPa. However, the predicted in-plane failure stress is 1807 MPa compared to the measured value of 679 MPa. Hence, delamination still contributed to the early failure of the specimens. The film adhesive layer was responsible for the delay in the delamination initiation load of the specimen and the consequent higher failure load of the specimens. The failed specimens shown in Fig. 6 exhibit some delamination but considerably more fiber breakage than the control specimens. The increase in fiber breakage is associated with the higher load achieved by the specimens at failure.

Delamination initiation can be fully suppressed by using a thicker layer of film adhesive. It is analytically determined that a 5-mm-thick film adhesive layer is necessary to suppress the delamination initiation completely. The resulting considerable reduction in the peak interlaminar shear stress σ_{1z} at the $+15^\circ/\text{FA}$ interface was compared to the $+15^\circ/-15^\circ$ interface of a control specimen is shown in Fig. 7. However, a thicker film adhesive strip will also cause higher interlaminar stresses in the drop-off region within the laminate [15]. Note that this somewhat unreasonable film adhesive thickness of 5 mm necessary to suppress delamination totally is a direct result of the large effective ply thickness used to create a large boundary layer. In actual applications, smaller effective ply thicknesses would be used, and considerably thinner film adhesive interlayers could be used to suppress delamination totally [9].

Specimens with 9-, 6-, and 3-mm-wide strips of film adhesive at the free edge showed a delamination behavior very similar to that of the specimens with a full-width ply of film adhesive. Their average delamination initiation stress was within 10% of that of the specimens with full-width plies of film adhesive. The average fracture stress for these specimens also falls within 10% of that for the specimens with full-width plies of film adhesive. Photographs of the failed specimens with film adhesive strips of 9-, 6-, and 3-mm width, shown in Fig. 8, indicate

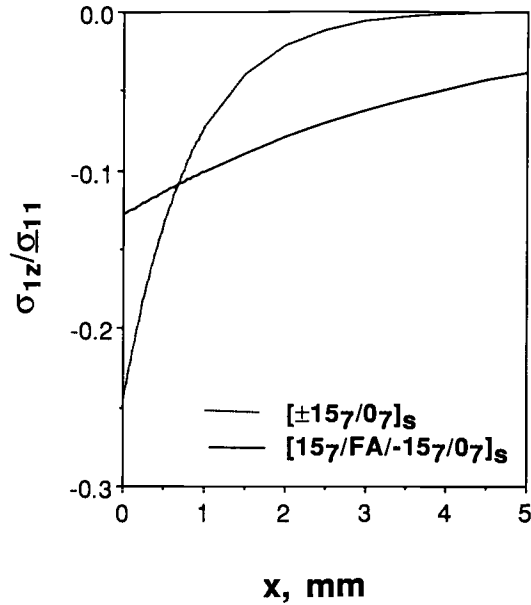


FIG. 7—Distribution of interlaminar shear stress, σ_{1z} , at critical interfaces of a $[+15^\circ/0^\circ]_s$ laminate and a $[+15^\circ/0^\circ]_s$ laminate with 5.0-mm-thick film adhesive strips.

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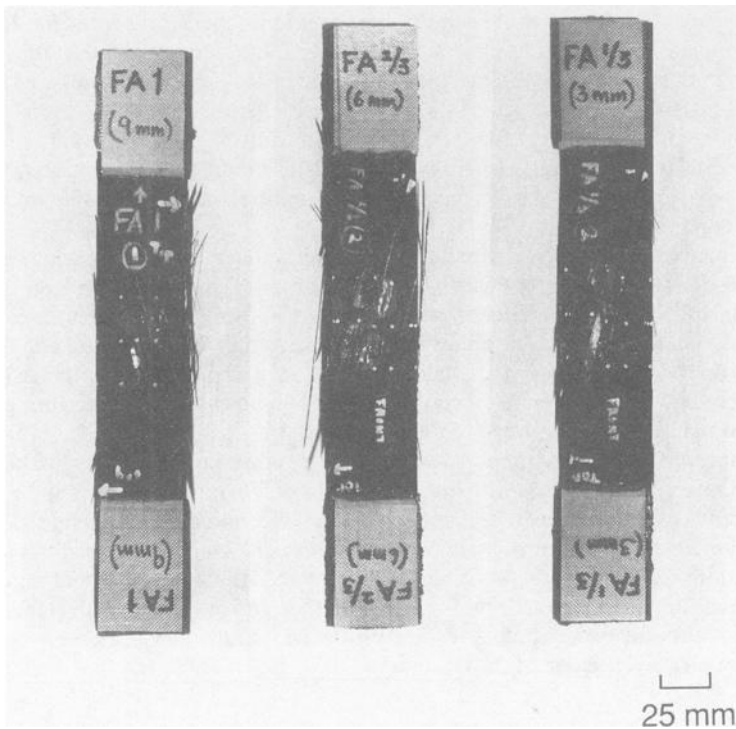


FIG. 8—Photograph of typical failures of $[\pm 15/0]_2$, specimens with film adhesive strip widths of (left) 9 mm, (center) 6 mm, and (right) 3 mm.

that these specimens exhibit failure modes consisting of delaminations and considerable amounts of fiber failure, again similar to that of specimens with full-width plies of film adhesive as shown in Fig. 6.

The results from the specimens with film adhesive indicate that the film adhesive placed at the delamination critical interface was successful in delaying the delamination initiation stress and, hence, increasing the failure stress of the laminate. Such interleaf layers can, in fact, totally suppress delamination if the proper thickness is used. Furthermore, since the differences in the delamination initiation and failure stresses of the specimens with varying widths of film adhesive seem to be insignificant, it can be said that all the widths of film adhesive that were manufactured were equally successful in delaying delamination initiation.

In the cases in which film adhesive strips of various widths have been used, there are two sets of interlaminar stresses that arise: the interlaminar stresses at the free edge and the interlaminar stresses arising where the film adhesive strips are terminated. These latter stresses, termed drop-off interlaminar stresses, could potentially cause delamination at the drop-off. They could, furthermore, interact with the stresses as a result of the free edge causing the delamination suppression capability to be diminished. This will depend on the size of the boundary layer associated with each set of stresses.

An analysis was developed to determine the stresses caused by such a drop-off. The results, detailed in Ref 15, indicate that the resulting interlaminar stresses at the drop-off are insignificant as compared to those at the free edge. This is attributed to the fact that the film adhesive, being thin and compliant, carries a very small fraction of the in-plane load. The boundary layer region at the drop-off extends to about 6 mm on either side of the drop-off. The free edge

boundary layer extends about 9 mm from the free edge. Thus, there is potential for the boundary layer regions to interact, resulting in varying levels of interference between the two types of interlaminar stresses. However, since the magnitude of the drop-off stresses are small, this interaction is insignificant in the current cases. This result is borne out by the fact that the delamination initiation stresses of the specimens with various widths of film adhesive are virtually the same as those of the specimens with full widths of film adhesive, implying that the drop-off stresses do not significantly affect the free edge interlaminar stresses involved in the delamination process.

The experimental results indicate that any manufacturable film adhesive width is capable of suppressing delamination in the cases in which the drop-off stresses are not large enough to cause a significant change in the interlaminar stress state in the vicinity of the free edge. It has been previously mentioned that the analysis showed that a 5.0-mm-thick film adhesive layer would be capable of completely suppressing delamination in a $[\pm 15_7/0_7]_s$ laminate. The drop-off stresses obtained using the analysis detailed in Ref. 15 indicate that the resulting drop-off stresses are comparable in magnitude to the σ_{2z} and σ_{zz} stresses at the free edge, which are an order of magnitude smaller than the σ_{1z} stresses at the free edge. Thus, the delamination behavior of the specimens in such a case with different widths of film adhesive layers at the free edge may be different as a result of the interference of the free edge and the drop-off boundary layers causing an altered stress state at the free edge. The free edge boundary layer for a specimen with a 5.0-mm-thick film adhesive strip at the $+15^\circ/-15^\circ$ interface is about 23 mm, and the drop-off stresses die out within a 10-mm region on either side of the drop-off. Hence, in such a case, the film adhesive strip would have to be wider so that the drop-off occurs outside the free edge boundary layer region of the laminate.

Delamination Growth

The specimens used to study the effect of the film adhesive on the growth of the delamination had film adhesive plies at the $+15^\circ/-15^\circ$ interface and Teflon strips of 9-, 6-, and 3-mm width at the $+15^\circ$ /film adhesive interface at the free edge of the laminate. Results in Table 4 show that the first two types of specimens failed at stresses which were within 10% of the failure stress of 486 MPa of the control specimens. The specimen with the 3-mm Teflon implant, however, failed at a stress of 591 MPa, which is over 20% higher than that of the control specimens and only 13% lower than that of specimens with full-width plies of film adhesive and no Teflon implant. Since delamination was preinitiated in these specimens, there is no "delamination initiation stress" mentioned for these specimens. The failed specimens with 9- and 6-mm Teflon implants shown in Fig. 9 indicate extensive delaminations and some fiber breakage. The failed specimen with 3-mm Teflon implants in Fig. 9 shows much more fiber breakage than the other two specimens of this category because of the higher load achieved.

The results from the specimens with the 6- and 9-mm-wide implants indicate that their fracture behavior is similar to that of the control specimens. Therefore, in the presence of a preinitiated delamination within the film adhesive width, the film adhesive is not capable of suppressing the growth of the delamination. However, the results from the specimens with 3-mm-wide implants vary from the other two. This implies the possible existence of a limiting delamination initiation size. In other words, when any undamaged specimen is tested, it exhibits a delamination initiation of a certain critical size at the delamination initiation load. This can be termed a "pop-in" delamination. Any preinitiated delamination of a size smaller than this critical size is not considered as an initiation by the laminate, and hence, the growth of such a delamination will not occur. The specimen continues to get loaded until the "pop-in" delamination initiation actually occurs and only subsequently does the growth take place. The pos-

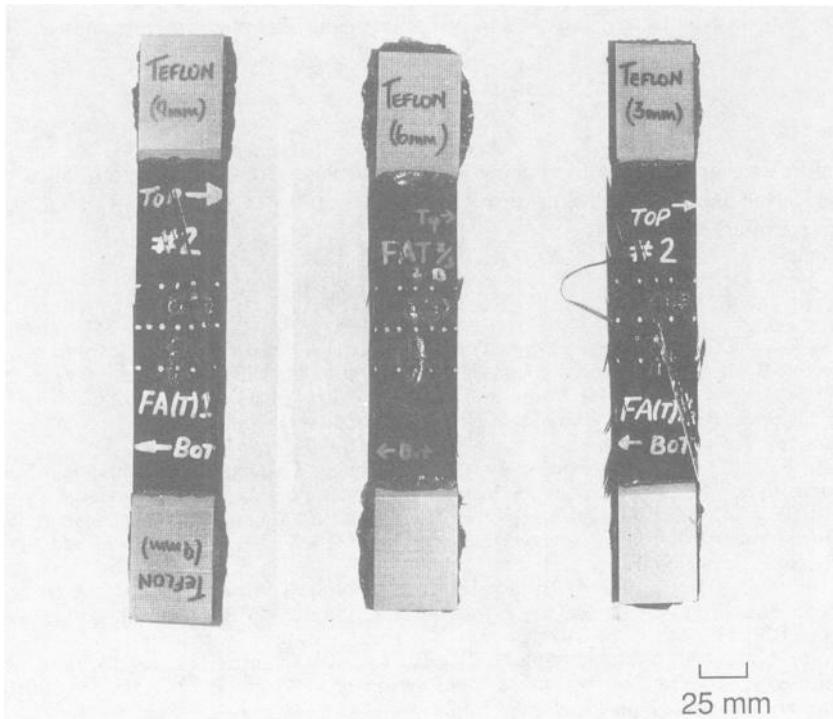


FIG. 9—Photograph of typical failures of $[157/FA/157/07]$, specimens with Teflon implants of width (left) 9 mm, (center) 6 mm, and (right) 3 mm.

sible existence of such a “critical size” of delamination initiation needs to be further investigated.

Summary

The use of film adhesive in the form of strips at critical interfaces restricted to the free edge of the specimens was proven to be a successful and efficient method for suppressing delamination. The layer of film adhesive is responsible for a reduced compliance mismatch and thus causes a reduced magnitude interlaminar stress state at the free edge which results in delaying the delamination initiation. The strips used in this study indicate that the suppression of delamination initiation by these strips is independent of the width of the film adhesive strip, if the stresses at the drop-off region of the strip are not significant compared to the free edge interlaminar stresses. If this is not so, then the boundary layer of the two must be separated to prevent their interference from altering the interlaminar stress state at the free edge.

It has also been shown that the film adhesive is only capable of delaying delamination initiation and not the growth of delamination within the strip width. Thus, in the presence of a preinitiated delamination, the film adhesive is not successful in suppressing the growth of the delamination. However, the results also suggest that a delamination initiation of a certain critical size is necessary for it to be “recognized” by the specimen as an initiation. It is therefore likely that an undamaged specimen on being loaded will exhibit a sudden pop-in delamination

above the critical size at the delamination initiation load. This phenomenon merits further investigation.

Acknowledgments

This effort was supported by the Boeing Military Airplane Company. The authors wish to acknowledge the assistance of John Masters, then of American Cyanamid, in providing the film adhesive interleaf material.

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Session VIII
Damage, Flaws, and Repair
Susan S. Avery, Chairman

Overview

In the aerospace industry, many of the design allowables for composite materials are based upon assuming the presence of damage or flaws. Of critical importance is the ability to (a) develop design concepts to maximize damage resistance and damage tolerance, (b) predict performance degradation with damage, (c) detect damage, and (d) effectively repair structure if appropriate. This session includes four papers which contribute to furthering our understanding in each of these four areas.

The first paper, "Damage Tolerance of Three-Dimensional Braided Commingled PEEK/Carbon Composites" by Hua, Chu, and Ko discusses a technique for reducing impact damage through the use of three-dimensional (3-D) braided composite structures. Improved impact damage resistance results in an effective increase in damage tolerance over two-dimensional laminate structures impacted under the same conditions. Open hole tensile and compressive residual strength is lower than that of a laminated structure, although the relative strength loss is better with 3-D braided structure.

Tan and Kim's paper "Damage Accumulation and Fracture of Notched Composite Laminates Under Tensile and Compressive Loading" is especially interesting because it includes compressive as well as tensile loadings. The characteristic length of damage at the tip of a notch is a critical parameter in many analytical models for predicting residual strength with damage. Tan and Kim show that the characteristic length of accumulated notch-tip damage is a constant for a given laminate if the damage characteristics are the same under different types of loading and varies if the damage characteristics vary.

Zhang and Sandor in "Advances in Thermographic Stress Analysis and Evaluation of Damage in Composites" present some experimental and analytical work on fibrous composite laminates. They show that, for laboratory specimens under cyclic loading, changes in the thermal signature in the region of the crack tip reflect damage accumulation. They developed a theory and method for calibrating changes in the thermal signature to detect and quantify damage growth in some monolithic composite laminates.

Finally, Siener in "Stress Field Sensitivity of a Composite Patch Repair as a Result of Varying Patch Thickness" presents his work on stress analysis of a variable thickness repair patch. Siener uses finite element modeling to obtain qualitative evaluations of the relative effectiveness of several patch designs and compares the results with experimental data. He concludes that a simple linear finite element modeling approach is inadequate for predicting the behavior of variable thickness repair patches. He also shows how plies reoriented into the principal stress direction can increase patch effectiveness and thus reduce patch thickness, except where the second principal stress is nearly as large as the first principal stress.

The technology necessary to establish acceptable levels of defects for the design and testing of composite structures that have such expected flaws and damage is provided by improving composite material damage tolerance through the use of 3-D braiding, monitoring damage accumulation and fracture of notched composites, and being able to detect, quantify, and predict damage growth by thermographic stress analysis. Once the damage occurs or the flaws are discovered, a method of designing repairs for them is presented.

Damage Tolerance of Three-Dimensional Commingled PEEK/Carbon Composites

REFERENCE: Hua, C. T., Chu, J.-N., and Ko, F. K., "Damage Tolerance of Three-Dimensional Braided Commingled PEEK/Carbon Composites," *Composite Materials: Testing and Design (Tenth Volume)*, ASTM STP 1120, Glenn C. Grimes, Ed., American Society for Testing and Materials, Philadelphia, 1992, pp. 400–413.

ABSTRACT. The concept of structural toughening of composites by three-dimensional (3-D) fiber architecture is demonstrated. Polyetherketone (PEEK) matrix in filament form was commingled with carbon fiber before preform fabrication and composite consolidation to facilitate matrix introduction into the 3-D fiber network. The effect of matrix viscosity was found to be essential for effective fiber wetting in the 3-D fiber network. The tolerance of 3-D composites to damage induced by open holes and by impact was investigated under tensile and compressive loading. 3-D reinforced composites were found to have higher compression after impact strength and an order of magnitude lower damage area under 13 J/mm impact energy in comparison to laminated PEEK/carbon composites. This high level of damage tolerance is due to the lack of delamination inherent to 3-D reinforced composites.

KEYWORDS: three-dimensional braid, angle-ply laminates, thermoplastics, commingled yarns, compression after impact strength, damage tolerance, failure mode, open hole compressive strength, open hole tensile strength

One of the key considerations in the design of composite aircraft structures is the ability of the composite to tolerate damage created during manufacturing (bolt holes, cutouts, and so forth) and damage inflicted while in service (bird strikes, dropped tools, runway debris, and so forth). Common modes of failure as a result of impact damage in laminated composites have been attributed to matrix cracking, delamination, and fiber breakage. The discontinuities or free edges created by cutouts in laminated structures also provide easy access to solvent and moisture, in turn increasing opportunities for delamination.

In addressing the poor damage tolerance problems inherent to laminated composites, the strategy has been to use tough or toughened materials and introduce through-the-thickness reinforcement. It was demonstrated that the use of thermoplastic matrices can produce an increase in composite fracture toughness as well as compression after impact strength compared to thermoset composites [1]. The introduction of through-the-thickness reinforcement by stitching in thermoplastic laminates, as demonstrated by Su [2], resulted in an increase in critical crack-opening load and a higher fatigue threshold as well as a fatigue life increase. The concept of structural toughening has been examined by a number of investigators [3–5]. Because there is no plane of lamination, it was found that three-dimensional (3-D) braided

Paper supported by the Army Research Office and Boeing.

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composites are more capable of resisting damage than laminated composites. In addition to its fully integrated nature, the 3-D braided structure was also recognized as being capable of net structural shape formation [6]. The commingling of thermoplastic with the reinforcing fiber has been demonstrated to be an effective approach in facilitating matrix infiltration into the dense 3-D fiber network [7]. The combination of thermoplastic matrix with 3-D braided structure appears to be an attractive structural toughening concept for the net shape manufacturing of structural components.

Accordingly, the objective of this study is to demonstrate the concept of structural toughening by 3-D fiber architecture in a thermoplastic matrix. The damage tolerance behavior of commingled polyetheretherketone (PEEK)/carbon 3-D braided composites was characterized by open hole tension (OHT), open hole compression (OHC), and compression after impact (CAI) testing. The damage response of 3-D braided composites was also compared to angle-ply PEEK/carbon composites with similar fiber orientation and fiber volume fraction.

Material System

The material systems used for this study were obtained from two previous studies [8,9] using commingled 380G PEEK/G30-500 carbon and 150G PEEK/AS4 carbon. The 380G PEEK is a thermoplastic matrix in filament form, with higher molecular weight (MW ~ 40 000) and higher viscosity (3800 centipoise at 400°C) than the 150G (MW ~ 23 000, 1500 centipoise) filament [10]. A summary of the typical properties of the PEEK matrix and carbon fibers is shown in Table 1 [11,12].

The benefits and methods of placement of thermoplastic matrices in filamentous form have been detailed elsewhere [7,8,13]. Intimately mixed with the carbon fibers in filament form, the PEEK/carbon commingled yarns have a fiber volume fraction of 60% for the 150G PEEK/AS4 and 50% for the 380G PEEK/G30-500, respectively.

With the matrix already mixed in the yarn bundle, the commingled yarns were fabricated into 3-D net shape braided preforms at the Fibrous Materials Research Laboratory at Drexel University. Figure 1 illustrates the incorporation of the commingled yarns into the 3-D braided structure and the net shape manufacturing possibilities with the 3-D braiding process. Detailed descriptions of the processing and structure of 3-D braided preforms have been presented in a review article [14].

The braided preforms represent two fiber reinforcement concepts consisting of 0 or 90° yarns in a 3-D braided fiber network. The 380G PEEK/carbon preform has 10% transverse (90°) lay-in, the 150G PEEK/carbon preform has 10% longitudinal lay-in yarns distributed uniformly in a 3-D braided structure. The 3-D braids have a surface braiding angle of ± 20°. Before consolidation, the 3-D braided preforms were cut into specific specimen lengths for each test as shown in Table 2.

To facilitate the handling of specimens, two methods were used to prevent the specimen ends from unbraiding. In one, the PEEK filaments were melted with a soldering gun before

TABLE 1—Properties of PEEK matrix and carbon fiber

Properties	PEEK	AS4-3K	G30-500
Tensile modulus, GPa	4.5	235	234
Tensile strength, MPa	69.0	3795	3792
Strain, %	4.7	1.53	1.62
Density, g/m ³	1.32	1.80	1.78
Melting point, °C	343		

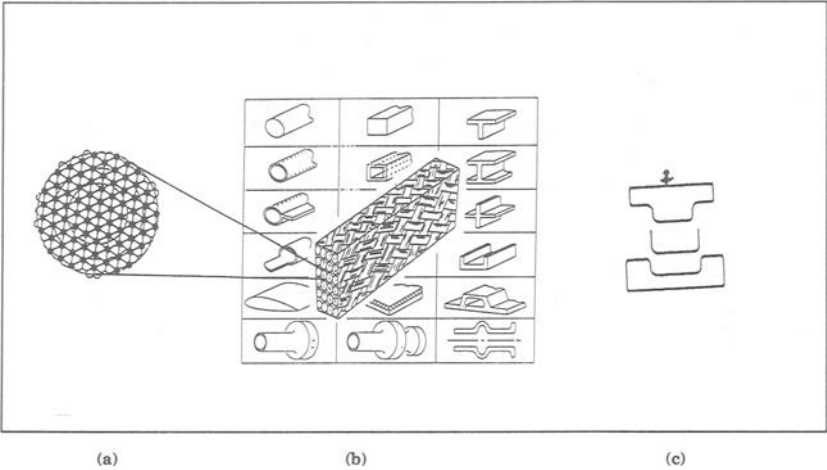


FIG. 1—(a) Schematic drawing of commingled yarn (b) net shape 3-D braiding, and (c) press forming.

cutting; in the other, the specimen ends were wrapped with Kapton® tape and cut through the Kapton.

After the 3-D preform was cut to size, it was placed in a match die steel mold. The steel mold was built with replaceable side rim bars, allowing width and length adjustment. A detailed description of the mold assembly has been given in Ref 8. Kapton film was wrapped around the specimen before placement to achieve a well-defined final net-shape composite. This prevented the formation of jagged edges caused by resin squeeze-out; therefore, no cutting was necessary along the sides of the specimen and only mild milling of the edges was required.

The consolidation cycle for the composite consisted of three stages: heat up, hold, and cool down. A maximum heat up rate of 5°C/min was achieved using a 200-ton (181-mg) Wabash press. The heat-up process took about 2 h to reach 400°C. A longer dwell time was required (compared to conventional unitape prepreg such as APC-2) to accommodate the dense 3-D commingled fiber network braid. The 3-D sample was held for 45 min before a 4°C/min cool down under pressure. The consolidation quality of the composite was evaluated with optical microscopy. As shown in Fig. 2, a higher degree of fiber wetting occurred in the 150G PEEK/carbon than in the 380G PEEK/carbon composite. In general, the quality of these specimens is very good; voids contained in these specimens are negligible.

For comparison purposes, 18-ply APC-2 unitape laminates with 60% fiber volume fraction and a stacking sequence of $[(\pm 20)_4, 0]_s$ were also fabricated and processed under similar processing conditions.

TABLE 2—Dimensions for 3-D braided preforms.

Specimen Type	Dimension, cm
Open hole tension	3.8 by 30.5 by 0.25
Open hole compression	3.8 by 30.5 by 0.25
Compression after impact	10.16 by 15.24 by 0.36

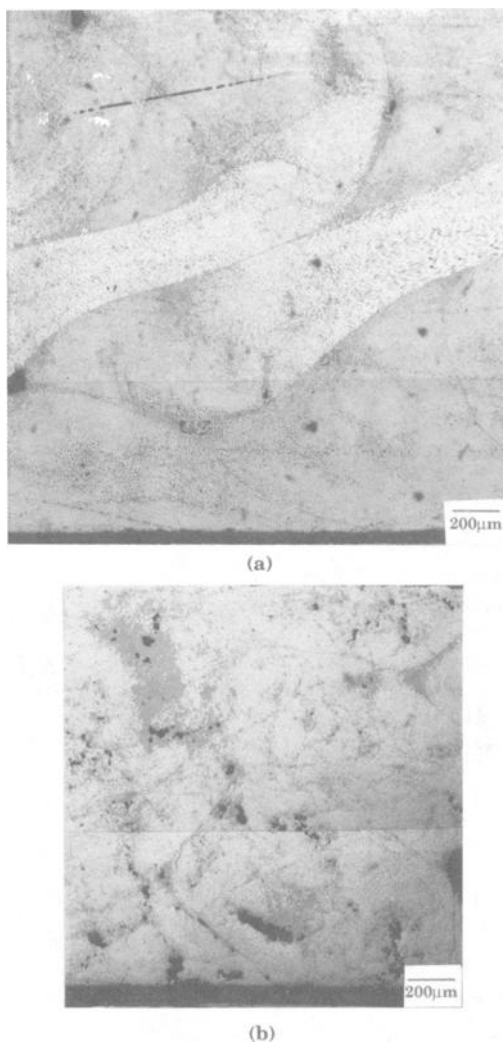


FIG. 2—Photomicrographs of cross-sectional areas of 3-D braided commingled PEEK/carbon composites: (a) 150G PEEK/carbon and (b) 380G PEEK/carbon.

Testing Methods

Testing of the composites was conducted according to Boeing Support Standard, Advanced Composite Compression Tests (BSS 7260) using open hole tension (OHT), open hole compression (OHC), and compression after impact (CAI) methods. Microscopic and macroscopic observations were also made to assess the effect of matrix viscosity on fiber wetting as well as the fracture behavior of the composites after being subjected to various types of damage.

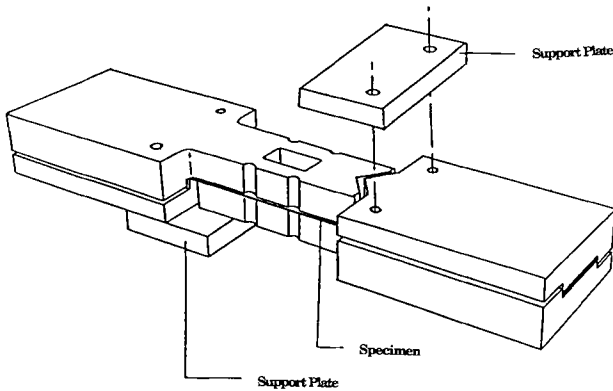


FIG. 3—Front elevation view of OHC test fixture with support plates.

Tension and Compression Test

The open hole (OH) specimens, measuring 38 by 305 mm (1.5 by 12 in.), were prepared by drilling a 6.35-mm ($\frac{1}{4}$ in.) hole at the center of the specimen. The OH specimens were tested on a servohydraulic mechanical testing unit under a constant strain rate of 1.27 mm/min (0.05 in./min). The specimens were tested in tension as well as compression. The OHC test was conducted with a support fixture as shown in Fig. 3, whereas the OHT test was done without a support fixture. In both cases, hydraulic grips were used. The OHT and OHC strength were calculated according to the following formula:

$$\text{OHT or OHC strength} = \frac{p}{bt} \quad (1)$$

where

- p = ultimate load,
- b = total specimen width (including hole), and
- t = specimen thickness.

Tension and compression tests for APC-2 laminated composites without holes were also conducted under the same testing conditions.

Compression After Impact Test

The CAI specimens were impacted under three levels of energy: 45, 57, and 68 J. A falling weight impactor with a 16-mm-diameter hemispherical tip was used. During the impact test, the specimens were clamped in a support base, as shown in Fig. 4a. After impact, the specimens were nondestructively inspected using a through transmission ultrasonic (TTU) scanner for damage assessment and the damage area of the specimens measured. Subsequent compression tests (test fixture shown in Fig. 4b) were performed according to Boeing Support Standard BSS 7260.

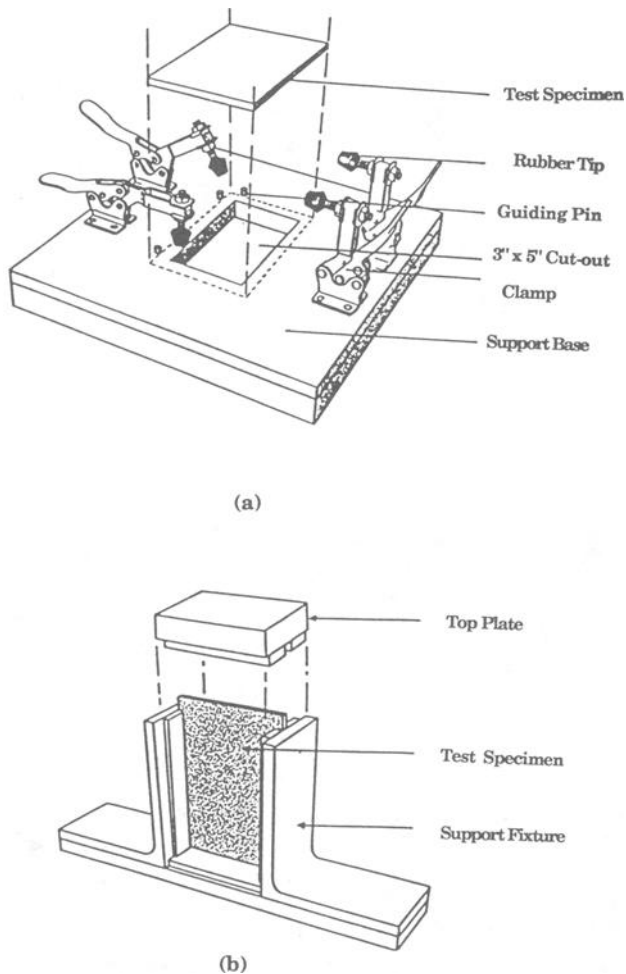


FIG. 4—(a) Impact test fixture and (b) CAI test fixture.

Results and Analysis

Response to Open Hole Damage

The response of the composites to open hole tension and compression is a measure of the sensitivity of the composite to notch or through-the-thickness defects. The OH responses are affected by fiber architecture and matrix viscosity. As expected, the compressive strength of the composite with or without hole is lower than tensile strength. Table 3 shows the tension, compression, and OHT and OHC strength of the laminated and 3-D braided composites. Each result was from three to five different specimens for 3-D braided and laminated composites (as a point of reference, the properties of a carbon/epoxy composite are also included). In general, one can see that the OHT and OHC strength of the laminated composites is higher than that of the braided composites for both thermoplastic and thermoset matrices. As illustrated by the

TABLE 3—Strength of thermoplastic and thermoset composites

	V_f , %	Tension			Compression		
		Without Hole, MPa	With Hole, MPa	% Strength Retention	Without Hole, MPa	With Hole, MPa	% Strength Retention
LAMINATE							
18-ply APC-2 [(±20°) ₄ ,0] _s	60	1081	628	58	646	363	56
16 ply APC-2/AS4 [−45/0/+45/90] _s [16,17]	60	910	411	45			
24 ply AS4/3501 [−45/0/+45/90] _s [3]	60	911	445	49	420	403	96
3-D BRAID							
150G PEEK/AS4 (0° ± 20°)	60	586	463	79		259	
380G PEEK/Celion (90° ± 20°)	50	538.9	277.4	52		190	
C12K/3501 (±20) [3]	60	668	661	99	429	314	73

laminated composites, OHT and OHC strength decreases with increasing fiber orientation angle. On the other hand, the level of strength retention for 3-D composites tends to be higher than that of laminates. Comparing the two 3-D braided thermoplastic composites (150G versus 380G), it can be seen that orienting fibers in the direction of loading coupled with a lower matrix viscosity resulted in higher OHT and OHC strength. An observation of the failure mode of the OHT and OHC specimens reveals that, although the laminated composites had higher strength, the 3-D braided composites contain damage better. In laminated composites, the predominant mode of failure is shear and delamination, whereas in the 3-D braided composites, failure is initiated by microdebonding and fiber breakage. As shown in Fig. 5, the OH laminate experienced matrix-dominated brittle mode of shear failure under compression. Similar modes of failure, as shown in Fig. 6, were also observed in the without hole specimens. Under tension, rupture of the uniaxial fiber occurs first followed by delamination. The lack of delamination in compression can be attributed to the nature of the OHC test wherein the specimen is supported to avoid buckling.

Fiber architecture in the 3-D braided composite was found to play a significant role in the damage resistance of the composite. For the 3-D braided composites having transverse (90°) fiber reinforcement, crack growth had a tendency to travel along the transverse direction in both tension and compression tests. Under tension, the crack travels along the braiding yarn for a short distance before propagating between the transverse yarns. Under compression, the crack tends to propagate between the transverse yarns, as shown in Fig. 7. Crack growth is further suppressed by compression. In general, the 3-D braided composites have a relatively more ductile mode of failure than laminated composites.

For 3-D braided composites with longitudinal reinforcements, fiber-dominated failure prevails. Under compression, failure tends to propagate transversely from the open hole. Fiber

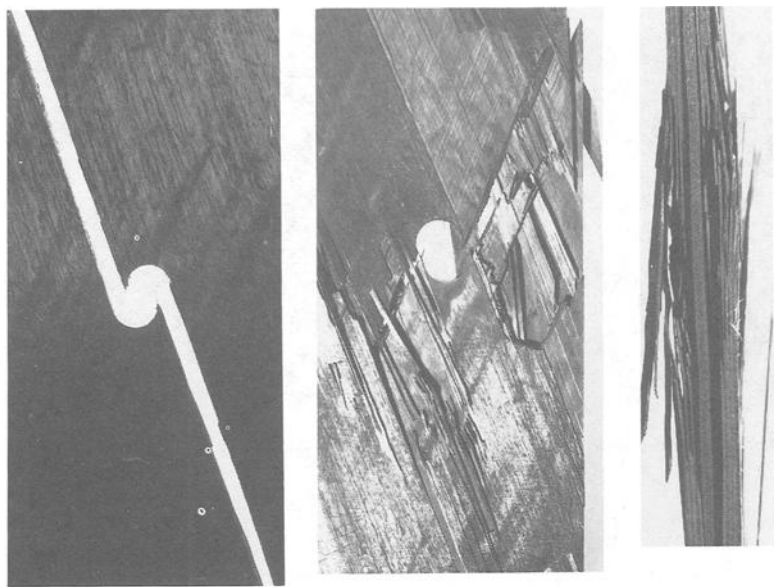


FIG. 5—Fracture behavior of laminated APC-2 composite: (a) OHC, (b) OHT, and (c) OHC (side view).

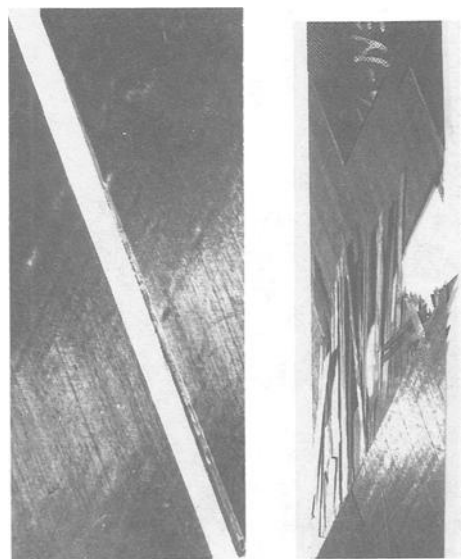


FIG. 6—Fracture behavior of laminated APC-2 composite: (a) compression and (b) tension.

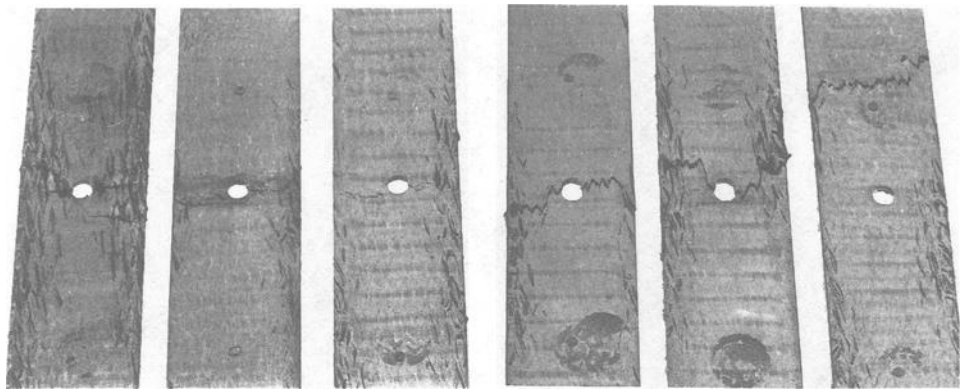


FIG. 7—Fracture behavior of 3-D braided 380G PEEK/carbon composite with transverse laid-in yarn: (a) OHC and (b) OHT.

breakage at yarn crossover gives a tough sawtooth type of failure, as shown in Fig. 8. In tension, failure occurs by microdebonding along the direction of the braiding angle.

Compression After Impact Response

The compressive strength of a composite after experiencing low velocity impact provides an indication of the specimen's built-up strength. It has been well established that delamination caused by low level impact can significantly reduce the compressive strength of laminated composites [18]. Accordingly, the compression after impact (CAI) response of the composites

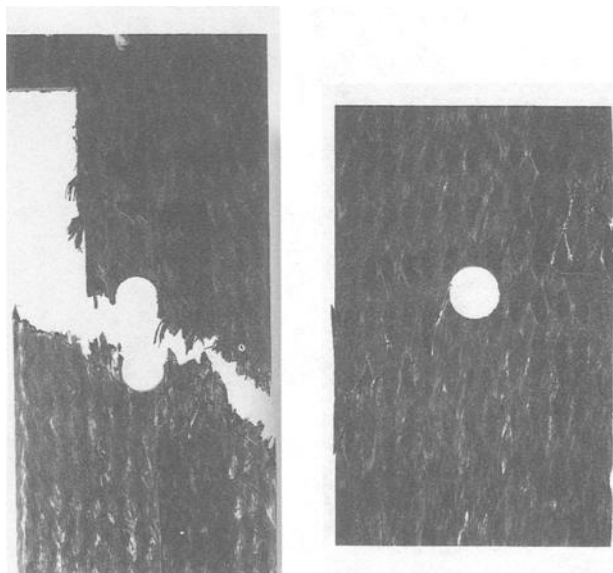


FIG. 8—Fracture behavior of 3-D braided 380G PEEK/carbon composite with longitudinal laid-in yarn: (a) OHC and (b) OHT.

were analyzed in terms of residual strength and damage or delamination area. Figure 9 compares the CAI strength of the 3-D braided PEEK/carbon composites with that of the laminated PEEK/carbon and epoxy/carbon composites. In general, the CAI strength of the composites decreases with increasing impact energy. It is quite apparent that the CAI strength of the thermoplastic composites is higher than that of the thermoset composites. In the family of PEEK/carbon composites, 3-D braided composites demonstrated a higher level of CAI strength.

Note the distinction of sensitivity to impact energy between 3-D braided 0 ± 20 and 90 ± 20 composites (this may be attributed to the difference in in-plane isotropy of the two composites). In composites having a 90 ± 20 fiber orientation, the fibers are less concentrated toward the longitudinal direction, resulting in a more balanced resistance to impact loading. On the other hand, in spite of the high level of CAI strength demonstrated by the 0 ± 20 composites at moderate level of impact energy (<10 g/mm), CAI strength decreases rapidly as impact energy level increases. It appears that a combination of 0/90 fiber reinforcement in a 3-D braided fiber network could lead to a composite with high residual strength and impact energy insensitivity.

The role of through-the-thickness reinforcement in damage containment can best be illustrated in the damage area-impact energy relationships shown in Fig. 10. It can be seen that over the range of impact energy (5 to 17 J/mm), the damage area of the 3-D composites is below 1000 mm² (much lower than that of the other thermoplastic and thermoset laminated composites).

Complex failure modes were observed for the 3-D braided CAI specimens. As illustrated in Figs. 11 and 12, the impact side of the specimen under transverse impact loading experienced compressive failure while the back side of the specimen experienced tensile failure. When compressive force was applied to the impacted samples, the respective damages were magnified. The impact and back sides of the specimens failed in a manner similar to the OHC and OHT specimens, respectively. As shown in the side views of the impacted specimens, no sign of delamination was observed in the 3-D composites.

For the 90 ± 20 3-D braided PEEK/carbon composites, crack propagation was stopped by the transverse reinforcement. For the 0 ± 20 PEEK/carbon composites, cracks tended to propagate along the braiding angle.

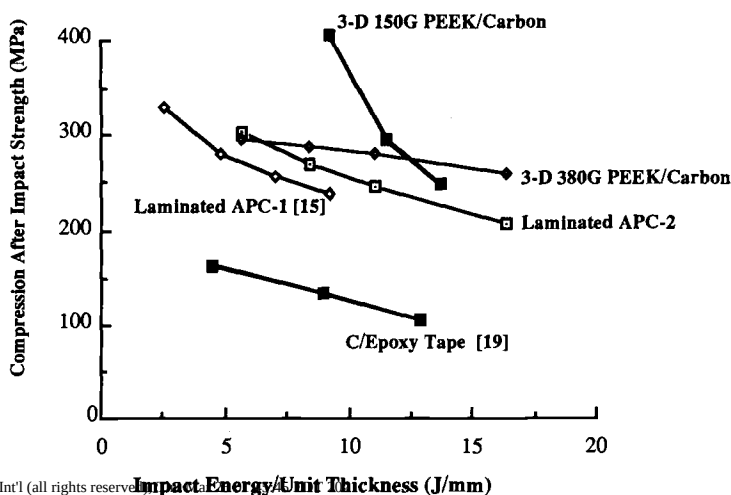


FIG. 9—CAI strength of composites at various impact energy levels.

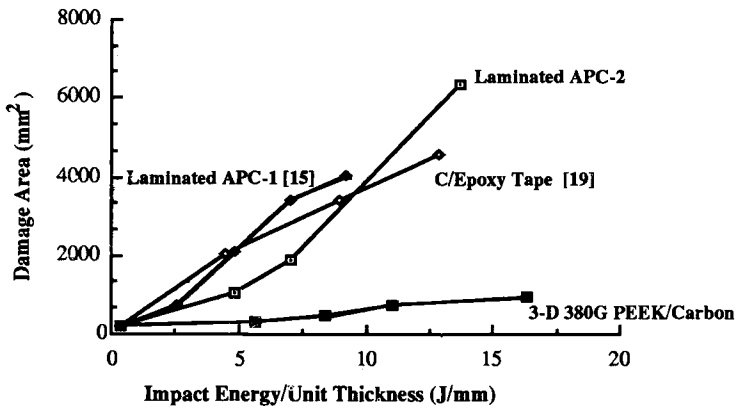


FIG. 10—The effect of impact energy on damage areas of 3-D braided and laminated composites.

Conclusions

The concept of damage containment through the use of 3-D fiber architecture was demonstrated in this study. Commingling a lower viscosity thermoplastic filamentous matrix with reinforcing fiber was shown to be an effective means of matrix placement in a 3-D dense fiber network.

The response of the composites to notch or through-the-thickness defects was characterized through OHT and OHC testing. Although OHT and OHC strengths of the laminated composites were higher than that of the 3-D braided composites, the intensity of damage propa-

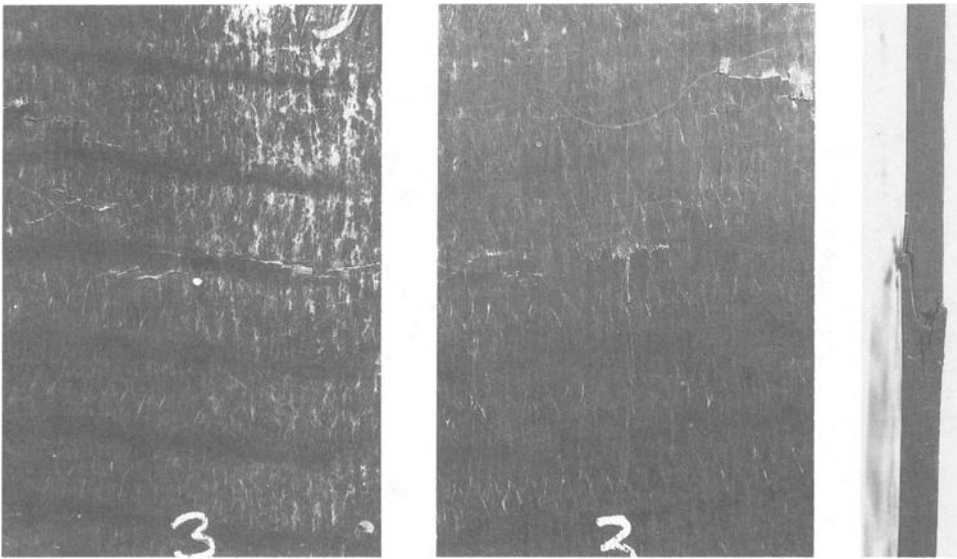


FIG. 11—CAI behavior of 3-D braided 380G PEEK/carbon composite with transverse laid-in yarn: (a) impact side view, (b) back view, and (c) side view

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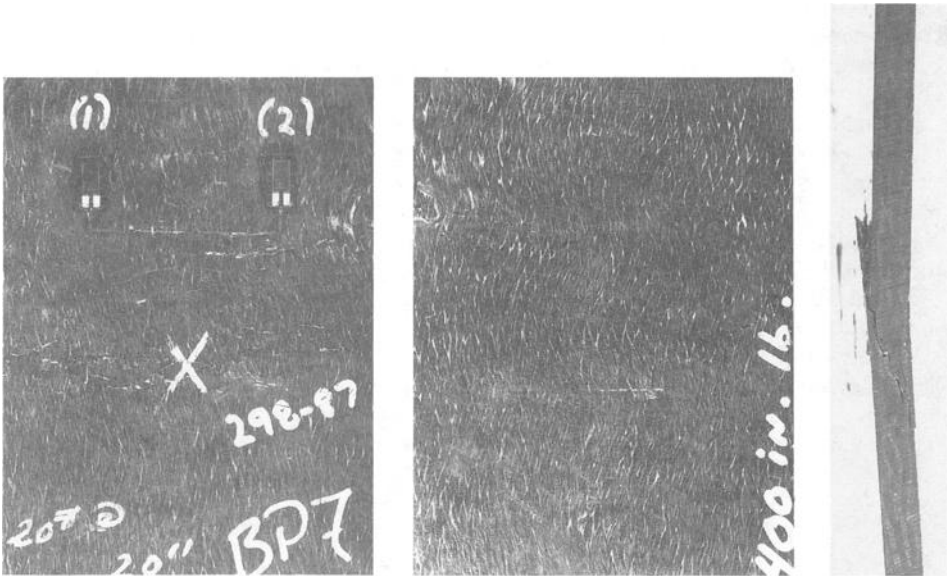
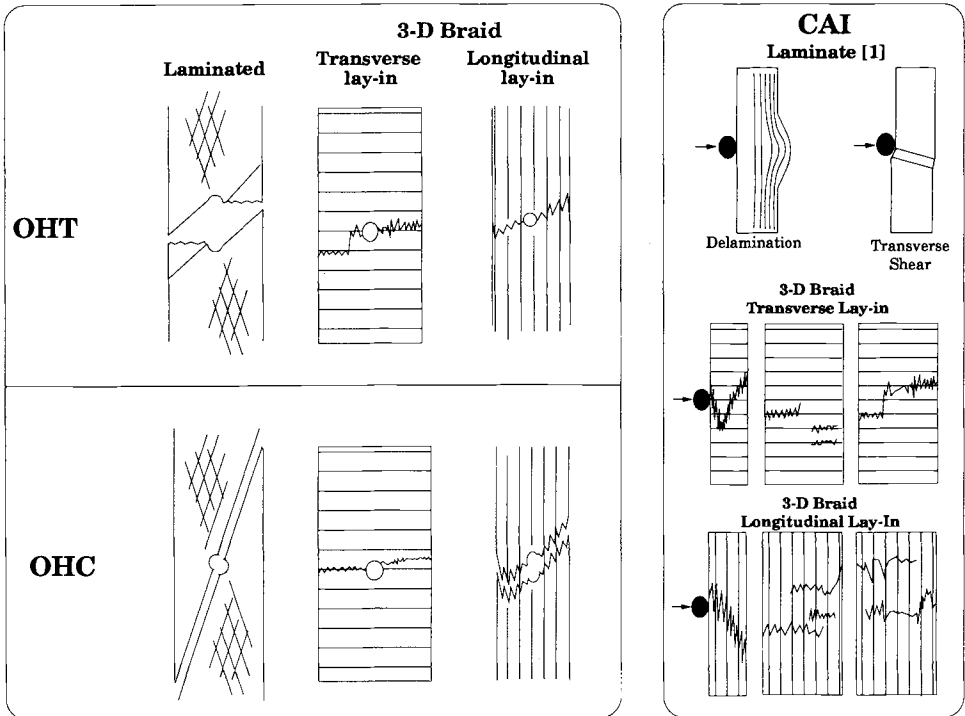


FIG. 12—CAI behavior of 3-D braided 150G PEEK/carbon composite with longitudinal laid-in yarn: (a) impact side view, (b) back view, and (c) side view.



gation within the 3-D composites was lower. The introduction of crack arrestors in the 3-D composites perpendicular to the direction of crack propagation provides additional damage containment. Further investigation is required to identify damage threshold and optimal distribution of these crack arrestors.

The response of the composites to impact damage was characterized by CAI testing. Because of impact damage eccentricity, the impact side of the specimens assumed a damage mode similar to that of an OHC specimen, whereas the back side of the specimen assumed a damage mode similar to that of an OHT specimen. The CAI strength of the 3-D braided composites was found to be higher than that of the laminated composites. The introduction of crack arrestors along the direction of testing enhances CAI strength, while the introduction of crack arrestors perpendicular to test direction reduces impact energy sensitivity. A schematic summary of the failure modes for laminated and 3-D braided composites is shown in Fig. 13.

As a result of through-thickness reinforcement, 3-D composites had a significantly lower level of damage area than laminates. With an ability to form complex structural shapes, 3-D braiding of commingled thermoplastic/carbon appears to be an attractive candidate for near-net shape manufacturing of high damage-resistant composites.

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Damage Accumulation and Fracture of Notched Composite Laminates Under Tensile and Compressive Loading

REFERENCE Tan, S. C. and Kim, R. Y., "Damage Accumulation and Fracture of Notched Composite Laminates Under Tensile and Compressive Loading," *Composite Materials Testing and Design (Tenth Volume)*, ASTM STP 1120, Glenn C. Grimes, Ed., American Society for Testing and Materials, Philadelphia, 1992, pp. 414-427

ABSTRACT: Damage accumulations before ultimate failure in laminated composites containing an open hole were studied and compared under tensile and compressive loadings. Four laminate layups, $[0_2/\pm 45]_{2s}$, $[0/90/\pm 45]_{2s}$, $[\pm 45]_{4s}$, and $[\pm 30/\pm 60]_{2s}$, were selected for the investigation. Two sets of laminates were prepared with one loaded in tension and another loaded in compression. They were examined using X-ray techniques after being loaded to specific load levels.

Damage accumulated in these laminates before ultimate failure can be correlated to the characteristic lengths used for the prediction of the notched strength. For laminates that the damage accumulations were found similar under different loading conditions, the same characteristic length can be used for the strength prediction under these loading conditions. On the other hand, if the damage mechanisms were different due to different type of loadings, then different characteristic lengths need to be used to predict the notched strength accurately. The experimental study presented in this paper reveals the physical explanation of using the characteristic length as a parameter for the theoretical strength prediction.

KEYWORDS: damage accumulation, composite laminates, characteristic length, minimum strength model, notched strength, X-ray techniques, tensile loading, compressive loading

Analytical models for the strength prediction of laminated composites containing stress concentrations include point stress and average stress criterion [1], point strength and minimum strength criteria in conjunction with a first-ply failure criterion [2], or a fiber-failure criterion [3], as well as some others documented in Ref. 4. The Whitney-Nuismer criterion [1] requires different characteristic lengths for a laminate subjected to different types of loadings [5,6]. To predict the notched strength of fiber-dominated laminates using Tan's models, Tan [2,3,7] and Tan and Kim [8] have shown that a same characteristic length can be used regardless of the loading conditions. For matrix-dominated laminates [9,10], the minimum strength model requires different characteristic lengths under different loading conditions.

As pointed out in an earlier work [3], the use of a characteristic length to predict the notched strength of laminated composites can be theorized and explained by the damage accumulation in the laminates. The damage accumulated in the laminates releases the stress concentrations. However, the magnitude of the characteristic length has not been correlated with the damage

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accumulation in the laminates. No experimental proof in the literature has shown whether the characteristic lengths and the damage accumulations should be the same or different under various types of loadings. This experimental study examines the damage accumulations of laminated composites subjected to tensile and compressive loadings. The result can verify if the characteristic length, which was used to predict the notched strength of the laminates, should be a function of the loading conditions.

Some experimental techniques, such as the photoelasticity, Moire methods [11], and holographic interferometry [12], can detect the surface cracks but not interior damage. The X-ray (radiographic) imaging technique can detect damage in both the exterior and the interior plies of a laminate. Therefore, this technique was applied to examine the extent of damage of a laminate after being loaded to certain load levels. The damage observed in a laminate subjected to tensile loading was compared to damage in laminate subjected to compressive loading at the same stress level and with the same specimen configuration. The damage mode and the extent of damage were then compared to the characteristic lengths used for the strength prediction.

Theoretical Background

The stress analysis of an infinite-width anisotropic laminate containing a central cutout has been derived using a complex variable method. Closed-form solutions were obtained in the case of a laminate containing an elliptical opening and subjected to in-plane loadings. The derivation was pursued in a superposition method. The laminate stresses (using index notation) along the 1-2 axes (Fig. 1) can be expressed as

$$\sigma_i = \sigma_i^* + \sigma_i^o, \quad i = 1, 2, 6 \quad (1)$$

where

- σ_i^* = the component as a result of the opening field and
- σ_i^o = the component as a result of the uniform stress field.

The solution of the σ_i^* component can be written in the following form:

$$\begin{aligned} \sigma_1^* &= 2\text{Re}[\mu_1^2\phi_1'(z_1) + \mu_2^2\phi_2'(z_2)] \\ \sigma_2^* &= 2\text{Re}[\phi_1'(z_1) + \phi_2'(z_2)] \\ \sigma_6^* &= -2\text{Re}[\mu_1\phi_1'(z_1) + \mu_2\phi_2'(z_2)] \end{aligned} \quad (2)$$

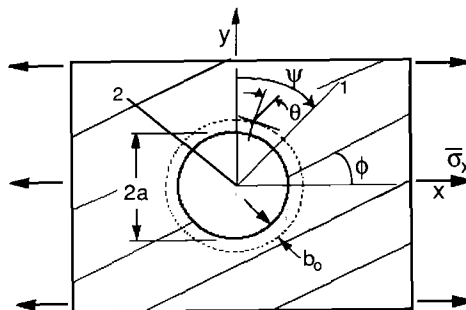


FIG. 1—An infinite-width laminated composite containing a circular cutout.

where

μ_1 and μ_2 = complex roots of a characteristic equation and
 ϕ_1 and ϕ_2 = complex potentials

The details of the derivation are given in Ref 7. Because all the specimens tested were finite-width, the fracture strengths were corrected using the finite-width correction factor [13] to obtain the notched strength for the infinite-width plate. Then the minimum strength model [7,9] was applied to predict the ultimate strength of the laminate. In this model, the strength, first-ply failure (FPF) or first-ply fiber failure (FPFF), is analyzed point-by-point and ply-by-ply along a characteristic curve given at a distance b_o from the opening contour (see Fig. 1). The characteristic curve is shown by the dotted line in Fig. 1. When the FPF criterion is applied, the strength reduction factor (SRF) of the laminate as a result of the hole is defined as

$$\text{SRF} = \frac{\sigma_N^\infty}{\sigma_o} = \frac{\text{FPF strength of the notched laminate}}{\text{FPF strength of the unnotched laminate}} \quad (3)$$

When the FPFF criterion is applied,

$$\text{SRF} = \frac{\sigma_N^\infty}{\sigma_o} = \frac{\text{FPFF strength of the notched laminate}}{\text{FPFF strength of the unnotched laminate}} \quad (4)$$

where σ_N^∞ and σ_o are the ultimate notched strength and the ultimate unnotched strength, respectively. Note that when the FPFF criterion is applied in conjunction with the minimum strength model, the values of both σ_N^∞ and σ_o can be predicted. Therefore, the only unknown is the characteristic length, b_o , which can be determined using one data point. Another way of determining the value of the characteristic length has been described in Ref 3.

Experimental Procedures

Laminated composite panels, 30.48 cm long by 30.48 cm wide (12 by 12 in.), were fabricated using Hercules graphite/epoxy AS4/3502 prepreg. All the notched and unnotched specimens were cut from the same panel. An earlier study [14] has shown that damages can be created in the specimen due to the machining process. Therefore, all the specimens were examined using X-ray techniques after the holes were cut using carbide tip drill bits. Only specimens that were free from damage were chosen to study the damage progression.

Four laminate layups, $[0_2/\pm 45]_{2s}$, $[0/90/\pm 45]_{2s}$, $[\pm 45]_{4s}$, and $[\pm 30/\pm 60]_{2s}$, were investigated. The first two laminates are fiber dominated, the third one is matrix dominated, and the last one is more difficult to be distinguished whether it is fiber or matrix dominated. Several combinations of hole-diameter/plate-width were fabricated (that is, $2a(\text{mm})/W(\text{mm}) = 5/37.9, 7.6/36.5, 15.5/38.1, 12.7/24.6$, and $15.4/25.5$ for the $[0_2/\pm 45]_{2s}$ laminates, $2a/W = 7.5/37.9, 12.7/30.6$, and $15.4/37.9$ for the $[\pm 45]_{4s}$ laminates, and $2a/W = 2.5/25.4, 5.1/38.1, 12.7/25.4$, and $15.5/25.4$ for the $[0/90/\pm 45]_{2s}$ and $[\pm 30/\pm 60]_{2s}$ laminates). All the specimens were 15.2 cm long with 2.5-cm endtabs on both ends.

Two sets of specimens were prepared for each layup. One was loaded in tension, and another was loaded in compression. For compression test, face-supported plates were used to prevent the specimens from buckling. The loading rates include 22, 44, and 88 N/s. These loading rates can be considered as quasistatic loading for high-strength materials. The slowest and the fastest rates were applied to the weakest and the strongest layups, respectively. It was expected that the ultimate strengths of the given four laminate layups would differ significantly. Therefore, different loading rates were used to minimize the dynamic effect but were intended to fracture

the specimens within a reasonable period of time. However, a constant loading rate has been used for a laminate configuration with any hole size or without a hole. After a specimen was loaded to a certain load level, it was unloaded and examined using X-ray techniques. To enhance the radiographic image, diiodomethane was applied around the hole and the free edges. The specimen was then reloaded to the next load level and examined again using X-ray after it was unloaded. This procedure was repeated until final fracture took place. We attempted to maintain the maximum load level for each cycle of compression at the same level as that under tension. This procedure enables one to compare the damage accumulation and damage mechanisms in a laminate caused by different types of loadings.

Results and Discussion

The experimental results include effective laminate properties and radiographs of the specimens.

Effective Laminate Properties

The load-strain relationships of the unnotched laminates were recorded using strain gages placed at the center of the specimens' surfaces. From these load-strain curves, we measured the initial moduli and Poisson's ratio. These values are regarded as the effective laminate properties. Using the tensile lamina properties given in Table 1, we calculated the effective laminate properties based on the classical laminated plate theory and compared them to the experimental result in Table 2. The predictions agree quite well with the experimental result. The experimental result reveals that compressive modulus of a laminate is consistently slightly lower than its tensile modulus.

Damage Accumulation and Mechanisms

The damage accumulation of the laminates has been studied using an X-ray technique. The extent of damage and the damage mechanisms of the laminates under compressive loading were compared to those of laminates subjected to tensile loading. In the following discussion, either or both the tensile and compressive specimens have been loaded to over 90% of their ultimate strength.

Figure 2 shows radiographs of two $[0_2/\pm 45]_{2s}$ specimens, 25.4 mm (1.00 in.) wide, containing a 15.4-mm (0.61-in.) diameter hole. Figures 2a and 2b were taken after the specimens were loaded to 216 MPa in tension and 209 MPa in compression, respectively. They both show fiber-matrix splitting at the edges of the opening in the 0° ply. A few matrix cracks, initiating

TABLE 1—*Tensile properties for the graphite/epoxy AS4/3502 lamina*

Parameter—Modulus and Poisson's Ratio	GPa (psi)	Parameter—Strength	MPa (ksi)
Longitudinal modulus, E_{11}	143.3 (20.79×10^6)	Longitudinal tensile strength, X	1862 (270)
Transverse modulus, E_{22}	10.7 (1.55×10^6)	Longitudinal compressive strength, X'	1655 (240)
In-plane shear modulus, G_{12}	6.1 (0.88×10^6)	Transverse tensile strength, Y	43 (6.2)
In-plane Poisson's ratio, ν_{12}	0.29	Transverse compressive strength, Y'	206.8 (30)
		In-plane shear strength, S	65 (9.4)

TABLE 2—Experimental^a and predicted effective laminate properties.

Layup	Parameter	Prediction, GPa (psi)	Experiment, GPa (psi)		Error, %
			Tension	Compression	
[0 ₂ /±45] _{2s}	<i>E_x</i>	83.1 (12.05 × 10 ⁶)	80.4 (11.7 × 10 ⁶)	73.6 (10.7 × 10 ⁶)	3.3
[0 ₂ /±45] _{2s}	<i>G_{xy}</i>	21.6 (3.14 × 10 ⁶)	...	...	...
[0 ₂ /±45] _{2s}	<i>ν_{xy}</i>	0.65	0.69	0.66	−4.7
[±45] _{4s}	<i>E_x</i>	21.1 (3.07 × 10 ⁶)	21.2 (3.08 × 10 ⁶)	19.9 (2.88 × 10 ⁶)	−0.5
[±45] _{4s}	<i>G_{xy}</i>	37.2 (5.39 × 10 ⁶)	...	...	...
[±45] _{4s}	<i>ν_{xy}</i>	0.74	0.75	0.73	−0.9
[0/90/±45] _{2s}	<i>E_x</i>	56.3 (8.17 × 10 ⁶)	52.7 (7.64 × 10 ⁶)	46.7 (6.77 × 10 ⁶)	6.9
[0/90/±45] _{2s}	<i>G_{xy}</i>	21.6 (3.14 × 10 ⁶)	...	...	...
[0/90/±45] _{2s}	<i>ν_{xy}</i>	0.30	0.32	...	−6.3
[±30/±60] _{2s}	<i>E_x</i>	41.2 (5.98 × 10 ⁶)	37.6 (5.45 × 10 ⁶)	36.3 (5.26 × 10 ⁶)	9.7
[±30/±60] _{2s}	<i>G_{xy}</i>	29.4 (4.26 × 10 ⁶)	...	...	...
[±30/±60] _{2s}	<i>ν_{xy}</i>	0.49	0.48	0.47	2.1

^a Average of two to three specimens.

from the hole boundary, were also observed near the middle of the two fiber-matrix splittings. These matrix cracks were caused by compressive stress when the specimen was loaded in tension and were caused by tensile stress when the specimen was loaded in compression. The damages appear to be quite similar under the two different loading conditions.

Radiographs of [0/90/±45]_{2s} specimens, 31.7 mm (1.25 in.) wide, containing a 12.6-mm (0.49-in.) diameter hole are illustrated in Fig. 3. Figures 3*a* and 3*b* are pictures of the specimens after being loaded to 227 MPa in tension and 229 MPa in compression, respectively. In the tension-loaded specimen, a significant amount of matrix crackings were found in the 90, +45, and −45° plies initiating from the hole boundary. The fiber-matrix splittings were also observed in the 0° ply at the hole boundary. No noticeable damages were found in the compression-loaded specimen.

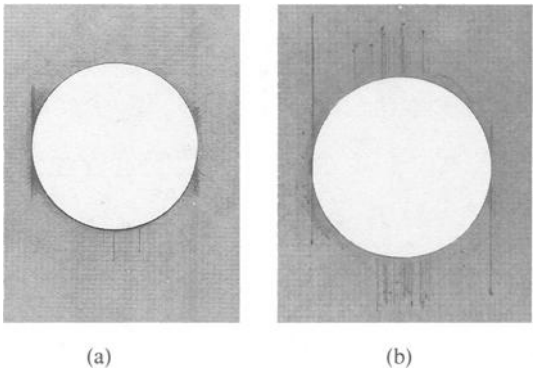


FIG. 2—Radiographs of [0₂/±45]_{2s} specimens, 2*a*/*W* = 15.4 mm/25.4 mm, after being loaded to: (a) 216 MPa in tension and (b) 209 MPa in compression.

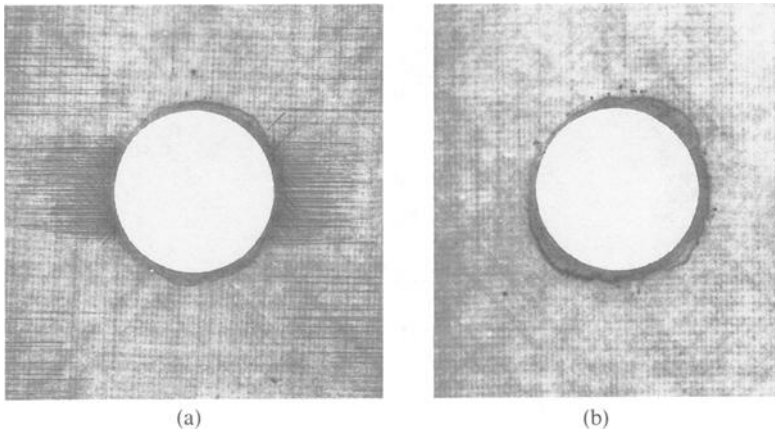


FIG. 3—Radiographs of $[0/90/\pm 45]_{2s}$ specimens, $2a/W = 12.6 \text{ mm}/31.7 \text{ mm}$, after being loaded to: (a) 227 MPa in tension and (b) 229 MPa in compression.

Figure 4 shows radiographs of $[\pm 45]_{2s}$ specimens, 31.6 mm (1.245 in.) wide, containing a 12.7-mm (0.499-in.) diameter hole. Figures 4a and 4b show the specimens after being loaded to 101 MPa in tension and 90 MPa in compression, respectively. The specimen under tensile loading (see Fig. 4a) exhibits significant matrix crackings in the 45 and -45° plies. The specimen under compressive loading (see Fig. 4b) has only a few tiny matrix cracks emanating from the hole boundary. The damage accumulated in the tension specimen is apparently much more than in the compression specimen.

Radiographs of $[\pm 30/\pm 60]_{2s}$ specimens, 25.1 mm (0.99 in.) wide, containing a 15.5-mm (0.61-in.) diameter hole are illustrated in Fig. 5. The pictures in Figs. 5a and 5b were taken after the specimens were loaded to 101 MPa in tension and 102 MPa in compression, respectively. Substantial matrix crackings were observed in the 60 and -60° plies initiating from the hole boundary of the tension-loaded specimen. Several matrix cracks also appear in the 60

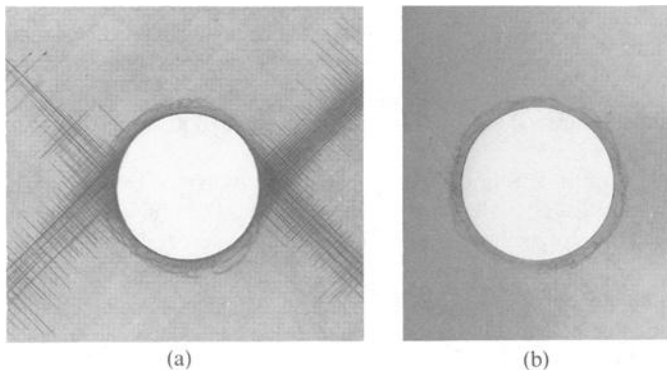


FIG. 4—Radiographs of $[\pm 45]_{4s}$ specimens, $2a/W = 12.7 \text{ mm}/31.6 \text{ mm}$, after being loaded to: (a) 101 MPa in tension and (b) 98 MPa in compression.

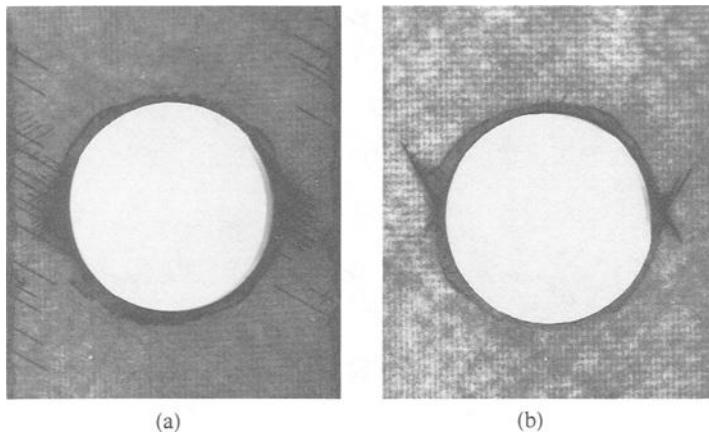


FIG. 5—Radiographs of $[\pm 30/\pm 60]_{2s}$ specimens, $2a/W = 15.5 \text{ mm}/25.1 \text{ mm}$, after being loaded to: (a) 101 MPa in tension and (b) 102 MPa in compression.

and -60° layers initiating from the straight edges. In the compression-loaded specimen, only four major matrix cracks were observed in the hole boundary. They all occur in the 30° and -30° plies.

Ultimate Strength

The fracture modes of the $[0_2/\pm 45]_{2s}$, $[\pm 45]_{4s}$, and $[\pm 30/\pm 60]_{2s}$ laminates under tensile and compressive loadings are illustrated in Figs. 6 through 8 and Figs. 9 through 11, respectively. The first layup (see Figs. 6 and 9) is apparently fiber dominated. No evidence of fiber breakage was found in the $[\pm 45]_{4s}$ laminates (see Figs. 7 and 10). The failure was dominated by matrix cracking. It is not easy to classify the $[\pm 30/\pm 60]_{2s}$ laminates (see Figs. 8 and 11) as matrix dominated or fiber dominated. Intuitively, this layup is matrix dominated. If we increase the angle of the $\pm 30^\circ$ plies, the failure would occur only in the matrix. However, if the angle of the $\pm 30^\circ$ plies decreases, fiber failure would be observed consistently. Therefore, it appears this layup lies between fiber- and matrix-dominated failure. Generally, the failure modes are similar for laminates containing holes of approximately the same size. However, when the hole size is reduced to a certain limit, the failure may not appear across the hole. The distinction between the tensile and the compressive failure modes is that the tension specimen fails under total separation, whereas the compression specimen does not separate into two pieces. The ultimate failure is defined when the specimen abruptly loses its load-carrying capability and cannot recover.

The gross strength of the specimens is the ultimate load divided by the total cross-sectional area of the specimens. Because all specimens have finite width, the gross strength was multiplied by the finite-width correction factor [13] to obtain the notched strength for infinite-width plates. The hole size effect can be shown by plotting the notched strength versus the hole radius. In the following comparison, each data point is the average of two to three results of replica specimens. Figure 12 shows the comparison of the predicted notched strength using the minimum strength model [3] with FPFF criterion and the experimental results for $[0_2/\pm 45]_{2s}$ laminates subjected to uniaxial tensile or compressive loading. The same characteristic length, $b_o = 0.56 \text{ mm}$ (0.022 in.), has been used for this layup under two different types of

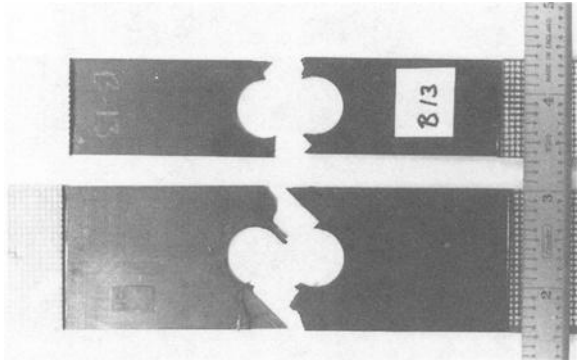


FIG. 6—Fracture specimens of $[0_2/\pm 45]_{2s}$ laminates subjected to tensile loading.

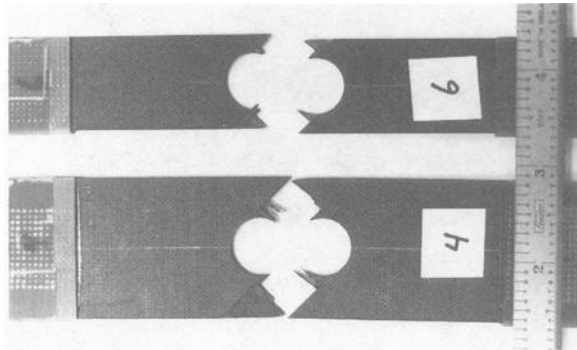


FIG. 7—Fracture specimens of $[\pm 45]_{4s}$ laminates subjected to tensile loading.

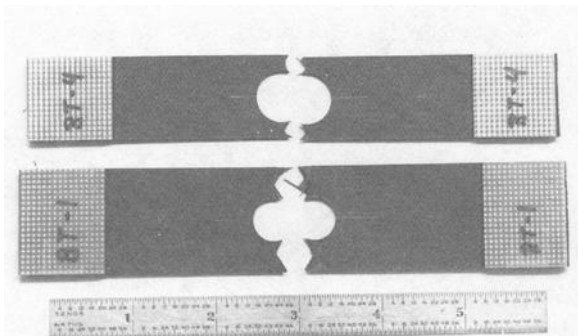


FIG. 8—Fracture specimens of $[\pm 30/\pm 60]_{2s}$ laminates subjected to tensile loading.

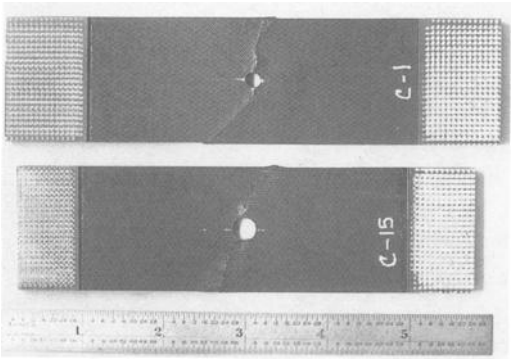


FIG. 9—Fracture specimens of $[0_2/\pm 45]_{2s}$ laminates subjected to compressive loading.

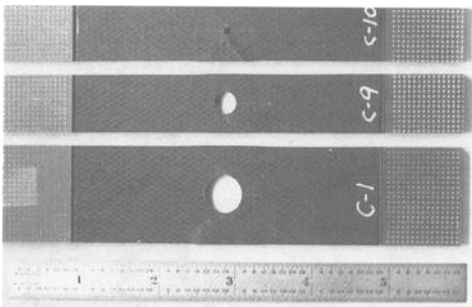


FIG. 10—Fracture specimens of $[\pm 45]_{4s}$ laminates subjected to compressive loading.

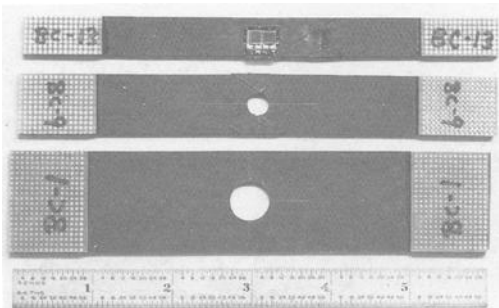


FIG. 11—Fracture specimens of $[\pm 30/\pm 60]_{2s}$ laminates subjected to compressive loading.

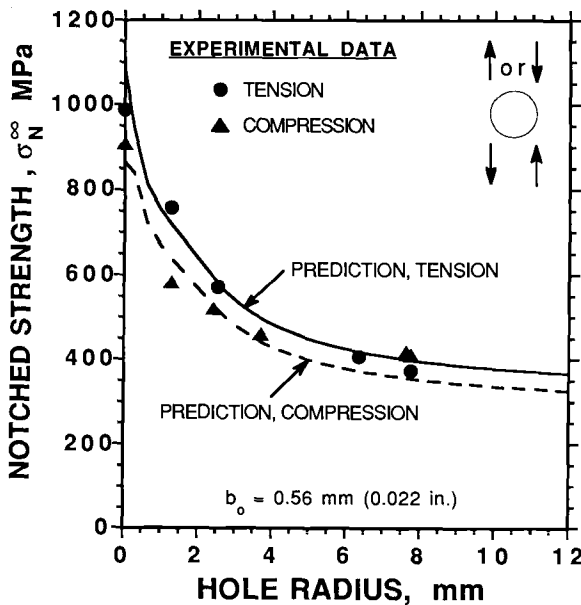


FIG. 12—Comparison of the predicted and the experimental notched strength for the $[0_2/\pm 45]_{2s}$ laminates.

For $[0/90/\pm 45]_{2s}$ laminates (Figs. 13 and 14), the characteristic length under tensile loading, $b_o = 0.46$ mm (0.018 in.), was found to be half of that under compressive loading, $b_o = 0.91$ mm (0.036 in.). The predicted notched strengths change much less than the magnitude of the characteristic lengths used for the prediction. For instance, when the hole radius is 3.81 mm, the predicted notched strength changes from 320 to 392 MPa (22.5%), whereas the characteristic length varies from 0.46 to 0.91 mm.

The correlations of the predicted and the experimental notched strength reduction factor for the $[\pm 45]_{4s}$ laminates are shown in Figs. 15 and 16 under tensile and compressive loading, respectively. Tsai-Wu failure criterion [15] has been used for the FPF analysis. The characteristic length used for the prediction of compressive strength, $b_o = 4.06$ mm (0.16 in.), is almost double of that for the prediction of tensile strength, $b_o = 2.54$ mm (0.10 in.). Although the value of the characteristic length changes from 2.54 to 4.06 mm, the variation of the predicted notched strengths is insignificant when the hole radius is less than 6.3 mm.

Figure 17 illustrates the result of the $[\pm 30/\pm 60]_{2s}$ laminates under tensile and compressive loading. Because this layup is in the boundary of fiber and matrix dominated, either FPF criterion or FPF criterion can be applied in conjunction with the minimum strength model. We have chosen the FPF criterion in Fig. 17. The characteristic length for the prediction under tensile loading, $b_o = 2.54$ mm (0.10 in.), is half of that under compressive loading, $b_o = 5.08$ mm (0.20 in.). The difference in characteristic length is similar to the prediction for the last two layups, $[0/90/\pm 45]_{2s}$ and $[\pm 45]_{4s}$.

Discussions and Conclusions

The damage accumulations and mechanisms of laminates containing a hole before ultimate failure have been studied as a function of the loading conditions (that is, tension versus com-

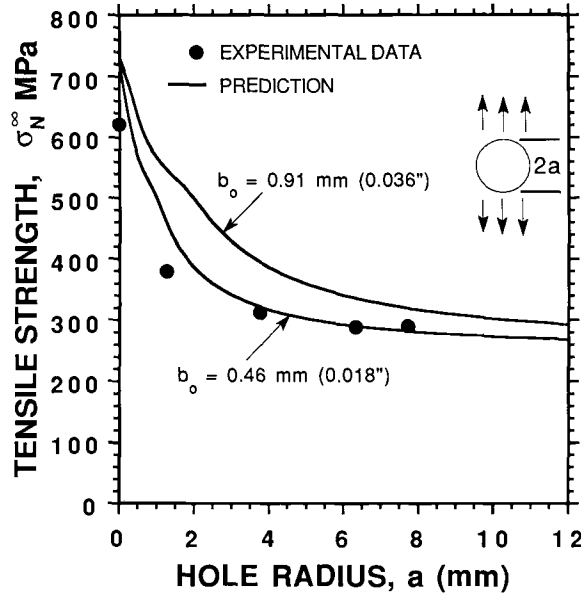


FIG. 13—Predicted and the experimental notched strength of the $[0/90/\pm 45]_{2s}$ laminates under tensile loading.

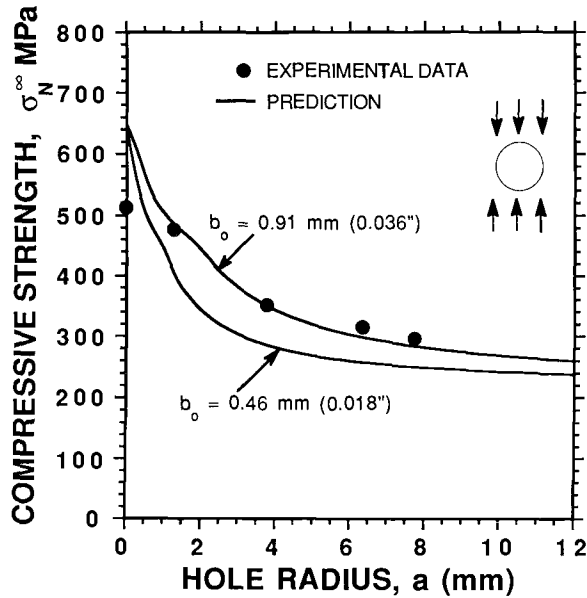
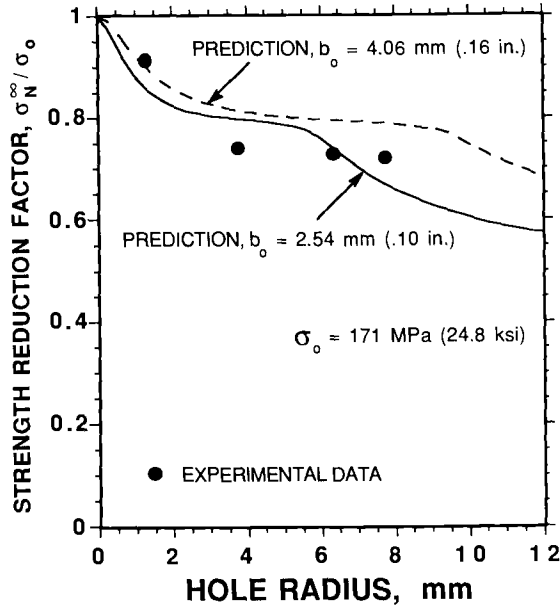
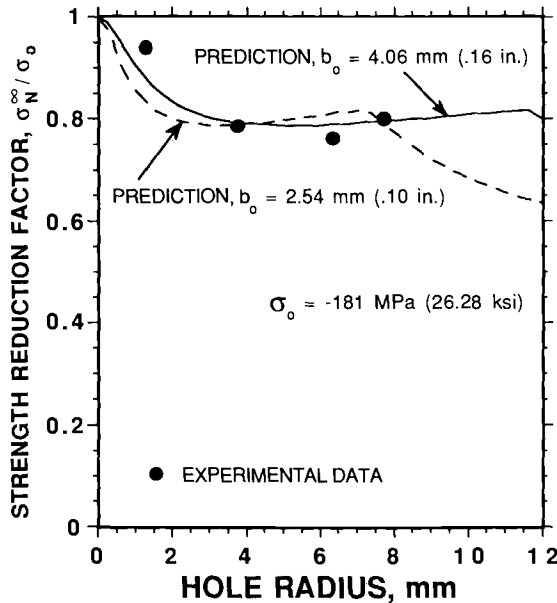


FIG. 14—Predicted and experimental notched strength of the $[0/90/\pm 45]_{2s}$ laminates under compressive loading.

FIG. 15—SRF of the $[\pm 45]_{4s}$ laminates subjected to tensile loading.FIG. 16—SRF of the $[\pm 45]_{4s}$ laminates subjected to compressive loading.

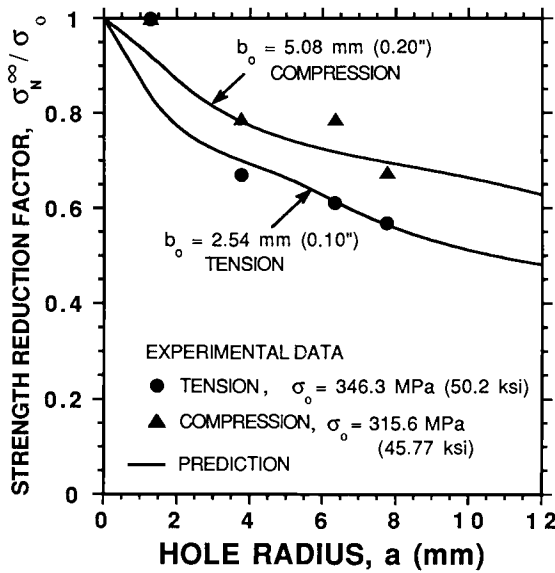


FIG. 17—SRF of the $[\pm 30/\pm 60]_{2s}$ laminates subjected to uniaxial tensile and uniaxial compressive loadings.

pression). The damage accumulated in the laminates can be related to the value of characteristic length, b_o , used for the strength prediction in the following.

1. For $[0_2/\pm 45]_{2s}$ laminates, the damage mechanisms before the ultimate failure are very similar under tensile and compressive loadings. Thus, the same characteristic length can be used to predict the notched strength of the laminates under different loading conditions accurately (see Fig. 12).
2. Before ultimate failure, the $[0/90/\pm 45]_{2s}$, $[\pm 45]_{4s}$, and $[\pm 30/\pm 60]_{2s}$ laminates have significant damage in the form of matrix cracks under tensile loading. However, much less damage was found in the laminates as a result of compressive loading (see Figs. 3 through 5). On the other hand, different characteristic lengths need to be used under different types of loadings to predict accurately the notched strength of the laminates. Therefore, the differences in damage accumulation and mechanism under various loading conditions result in different characteristic length used for the prediction.
3. For matrix-dominated laminates, the characteristic lengths are generally different because of different type of loadings.
4. The variation of the predicted notched strength is much smaller than the changes of the magnitude for the characteristic lengths used for the analysis.

Figure 13 shows that although there is less damage accumulated in a specimen under compressive loading than in a specimen under tensile loading, the characteristic length, b_o , used for the prediction is larger under compressive loading. A possible explanation is that the load-carrying mechanisms for a laminate with micro-damages are different under different loading conditions. Further study is needed for in-depth understanding. Finally, the prediction using a constant value of characteristic length appears to agree well with most experimental results,

although better agreement can be obtained if b_o is expressed as a function of the hole size. More experimental proof is needed to support the validity of this application of expressing b_o as a function of the hole size.

Acknowledgment

The authors wish to thank R. Esterline of the University of Dayton Research Institute for helping in preparation and testing of the specimens.

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Advances in Thermographic Stress Analysis and Evaluation of Damage in Composites

REFERENCE: Zhang, D. and Sandor, B. I., "Advances in Thermographic Stress Analysis and Evaluation of Damage in Composites," *Composite Materials Testing and Design (Tenth Volume)*, ASTM STP 1120, Glenn C. Grimes, Ed., American Society for Testing and Materials, Philadelphia, 1992, pp. 428–443.

ABSTRACT. Development of the thermographic stress analysis (TSA) method and the stress pattern analysis by thermal emission (SPATE) method for stress analysis and evaluation of damage in fiber-reinforced composites during the past five years is reviewed, and some new results are presented. The fundamental factors affecting the TSA signals, calibration methods, stress analysis, stress (strain) separation, and methods of qualitative and quantitative evaluation of damage are discussed. Reasonable future research topics are outlined.

KEYWORDS. fiber-reinforced composites, thermoelasticity, thermographic stress analysis, fatigue, damage evaluation, SPATE

Lord Kelvin in 1853 established a relationship between temperature change and deformation in an elastic solid [1]. Compton and Webster in 1915 verified Kelvin's theory using highly sensitive contacting resistance thermometers [2]. In the mid-1960s, Dillon and Taucher [3] applied the thermoelastic effect for the measurement of dynamic stresses with thermocouples. Belgen [4,5] systematically studied the thermoelastic effect for stress analysis by using infrared radiometry. Thermographic measurement techniques were applied in the 1970s to detect fatigue damage in composites [6,7]. In 1978, Jordan and Sandor [8] used differential temperature measurement for monitoring the mechanical behavior of materials. Following recent developments in thermographic techniques [9], quantitative stress measurements became possible by measuring the change in infrared photon emission produced by dynamic mechanical deformation. The TSA method has been used in isotropic materials for stress analysis, fatigue crack growth detection, predictions of stress intensity factors, and crack growth rate and residual stress measurements [10–14]. Compared to the numerous applications of the TSA method involving isotropic solids, the practical problems solved in composites are limited. There are several reasons for the few applications of the TSA method to composites. For example, the TSA maps for composites are not as smooth as for isotropic materials, as a result of microscopic and macroscopic inhomogeneities, practical calibration methods for stress analysis have been scarce, and two different thermal expansion coefficients are typically involved, leading to difficulty in interpreting the data.

In spite of the obvious difficulties, significant progress has been made in this area during the past five years. This paper reviews the current state of knowledge concerning applications of the TSA method to composites. Effects of thermal expansion coefficients, surface coating thickness, frequency, absolute temperature increases, and mean stress on the TSA signals are

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discussed. Calibration methods are evaluated, and a practical and effective method is found by using a unidirectional composite to obtain a required calibration factor and a material constant. Damage evolution tracked by the TSA method for notched specimens and bolted lap-joint laminates subjected to cyclic loads is discussed. Quantitative evaluation of damage in anisotropic materials using our recently developed thermoelasticity theory is presented.

Fundamental Factors Affecting TSA Signals

Classical Thermoelasticity Theory

The classical thermoelasticity theory gives an expression for the temperature change ΔT caused by a cyclic stress in a linear, homogeneous, and isotropic elastic material as

$$\Delta T = - \left(\frac{\alpha}{\rho C_p} \right) T \Delta I_1 \quad (1)$$

where

α = the thermal expansion coefficient,

C_p = the specific heat at constant pressure,

ρ = the mass density,

T = the absolute temperature, and

ΔI_1 = the cyclic range of the first stress invariant

Biot's work in the 1950s [15] is the earliest development of thermoelasticity theory for anisotropic properties. Until the early 1980s, no further work was carried out following Biot's study for anisotropic materials, possibly because of the complex constituents of the materials and the lack of adequate measurement technology. In 1986, Stanley and Chan [16,17] derived an explicit expression for an orthotropic material from the basic thermoelastic relationship between the stress changes and the temperature change as follows

$$\Delta T = - \left(\frac{T}{\rho C_p} \right) (\alpha_1 \Delta \sigma_1 + \alpha_2 \Delta \sigma_2) \quad (2)$$

where

α_1 and α_2 = the coefficients of thermal expansion in the directions of the principal material symmetry axes and

$\Delta \sigma_1$ and $\Delta \sigma_2$ = the stress changes along these directions

Effects of Thermal Expansion Coefficients

The TSA signal S recorded by a sensitive and linear detecting system is the change $\Delta \Phi$ in the infrared emission Φ resulting from a small temperature change ΔT at the surface of a cyclically loaded specimen. For broad-band photon flux detection [18],

$$S = \Delta \Phi = 3eBT^2 \Delta T \quad (3)$$

where

e = the surface emissivity and

B = the Stefan-Boltzmann constant

For an orthotropic composite under plane-stress conditions, S can be correlated with stresses by substituting Eq 2 into Eq 3

$$S = -3B(eT^3) \left(\frac{1}{\rho C_p} \right) (\alpha_1 \Delta\sigma_1 + \alpha_2 \Delta\sigma_2) \quad (4)$$

Note that the signal S not only is a function of stresses but also depends on the signs of the thermal expansion coefficients α_1 and α_2 . Some composite materials have positive values of both α (for example, glass fiber/epoxy composites), in which case the signals depend distinctly on the magnitude of the stresses. However, a unidirectional composite made of carbon fibers ($-\alpha$) and epoxy resin ($+\alpha$) has a net negative α for a fiber density above 0.5 in the direction parallel to the fibers and a positive α transverse to the fibers at room temperature. For instance, α_1 equals $-0.5 \times 10^{-7} K^{-1}$ and $-2.0 \times 10^{-7} K^{-1}$ at room temperature for fiber volumes of 52 and 61%, respectively, while α_2 equals 3.0 and $2.7 \times 10^{-5} K^{-1}$ [19,20]. Kageyama et al. [21] used this property for stress analysis in notched specimens. For a carbon/epoxy lamina, they reported $\alpha_1 = -0.4 \times 10^{-6} K^{-1}$ and $\alpha_2 = 3.0 \times 10^{-5} K^{-1}$. Because α_1 is relatively small here, the influence of $\alpha_1 \Delta\sigma_1$ on the signal S can be neglected for determining the stress $\Delta\sigma_2$ in the transverse direction.

In general cases, the TSA signal is the combination of the temperature changes resulting from the longitudinal and transverse stresses. The temperature change in the fiber direction is low (positive or negative) as a result of a very small (positive or negative) coefficient of thermal expansion in carbon composites, but the temperature change is large in the transverse direction because of the relatively large positive transverse coefficient of thermal expansion. However, the highly different thermoelastic properties of the carbon fiber and epoxy matrix as well as nonuniform fiber densities result in nonuniform infrared emission [17,22]. For example, considering a unidirectional carbon/epoxy composite under uniaxial loading ($+\Delta\sigma_1$, $\Delta\sigma_2 = 0$), the signal S (Eq 4) may be positive ($-\alpha_1$) in one location, or negative in other locations ($+\alpha_1$) because of nonuniform fiber densities.

Effects of the Surface Resin Thickness

Composite materials normally have a thin layer of pure matrix, for example, pure epoxy, at the surface. It has been reported [23,24] that the thickness of the surface layer of resin in carbon/epoxy composites has a substantial effect on the TSA signals because of the effect of thermal expansion coefficients (pure matrix $+\alpha$, carbon fiber $-\alpha$). The TSA signals increase linearly as the thickness of the resin layer increases from 0.1 to 0.6 mm. But the investigation in Ref. 24 has shown that when an originally 0.02-mm-thick surface resin layer (commercially manufactured) was removed entirely, significant differences in the TSA signals were found. On the other hand, when an additional 0.03- to 0.3-mm-thick epoxy resin was painted on the original manufactured epoxy surface, the TSA signals had an approximately constant value. The possible reason for this is that above a certain thickness the heat transfer is attenuated. More theoretical and analytical work is needed to understand this behavior.

Effects of Frequency

Equation 4, the expression for temperature change at the surface of a material, was derived for adiabatic condition. This means that heat-leakage effects are not covered by this equation. Note that, if the cyclic loading frequency is not high enough, internal conduction of heat between regions of different stress levels and heat loss from the surface to the environment may significantly affect the surface temperature [25]. Belgen [5] carried out experiments with

beams in bending to investigate the attenuation of thermoelastic signals as a result of heat leakage. He pointed out that, if a plate is subjected to out-of-plane bending, heat tends to leak from the side under compressive stress to that under tension. McKelvie [25] presented analytical methods for both internal and external heat transfer effects and stated that significant signal attenuation may occur under typical operating conditions as a result of internal conduction. The study in Refs 22 and 24 showed that adiabatic condition in composites is more difficult to ensure than in isotropic materials. The reported nearly adiabatic loading frequency is above 2 Hz in steel, above 5 Hz in aluminum, but above 30 Hz in composites [22]. It is apparent that the nonadiabatic behavior in composites is affected mainly by nonuniform fiber density, sub-surface heat conduction, and bithermal expansion coefficients (when applicable).

Effects of Absolute Temperature Change and Mean Stress

It is desirable to use as high a frequency as possible to ensure adiabatic condition. However, if a high-loading frequency is used, there is significant internal heating of a composite specimen. It was reported [24] that when the temperature increases from 306 to 337K, the signals detected on the surface of a carbon/epoxy specimen tested in a shaker at 150 Hz increase steadily from 250 to 1600 (uncalibrated stress units) with otherwise identical test conditions. This behavior is understandable considering Eq 4 and the fact that the surface emissivity also increases with temperature [24].

Recent work of Machin et al. [26] has shown that the TSA signal depends not only on the cyclic stress change but also on the mean stress applied. Wong et al. [27] proved that this mean stress effect is due to the temperature-dependence of the elastic properties of the material. The study of the mean stress effect serves two main purposes: to determine the calibration factor for different mean stresses, and to measure residual stresses in some cases [28]. Considering composites, Potter and Greaves [23] started from the first law of thermodynamics and obtained an expression that contains temperature-dependent terms. The experiments conducted on glass/epoxy and graphite/epoxy composites showed that the mean stress effects do exist and the detector signals increase with increasing mean stress and mean strain [24,28].

Calibration Methods and Stress Analysis

To verify Eq 4, Stanley and Chan [16,17] used four thin-walled Kevlar fiber/resin cylinders to determine the value of α_2/α_1 . The overall dimensions of each cylinder were: inner diameter, $d = 100$ mm, thickness, $t = 1.45$ mm, and length, $L = 32$ mm. The Kevlar fibers were wound in a balanced symmetric manner to give an orthotropic construction. Two different loadings, cyclic axial and internal pressure loads, were applied separately to each cylinder. When a cylinder was subjected to a cyclic axial compressive load of amplitude ΔF , the stress change in the axial direction was $\Delta\sigma_2 = -\Delta F/\pi d t$. Under a cyclic internal pressure loading of amplitude ΔP , the stress change in the circumferential direction was $\Delta\sigma_1 = -\Delta P/2t$. Substituting these stress changes into Eq 4, the ratio of α_2/α_1 was expressed as

$$K_2 = \frac{\alpha_2}{\alpha_1} = -\frac{S_y}{S_x} \frac{\pi d^2 \Delta P}{2 \Delta F} \quad (5)$$

where S_y and S_x are the TSA signals for axial and internal pressure loadings, respectively, with the same experimental and scanning details. Therefore, K_2 can be determined from the thermographic data and the applied stresses.

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Potter [29] derived an expression that relates S to the mechanical and thermal properties of

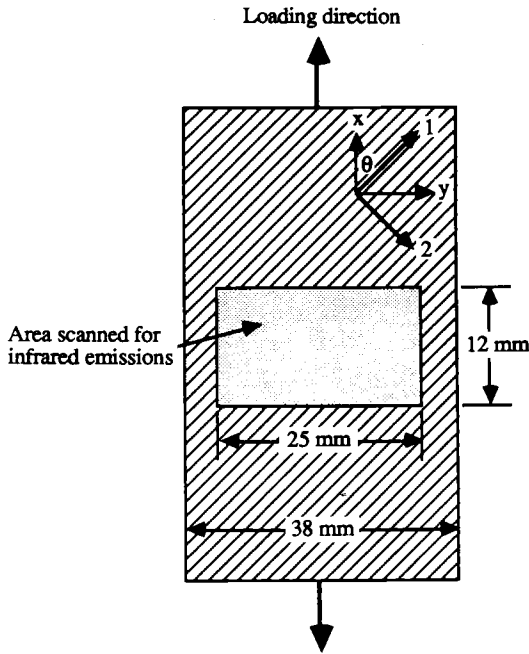


FIG. 1—Definition of axes and area scanned for TSA.

the surface ply and the laminate strains. If the surface ply fibers are parallel to the x axis as shown in Fig. 1, S can be expressed as

$$S = K_1[(Q_{11} + K_2Q_{12}) - v_{xy}(Q_{12} + K_2Q_{22})] \Delta\epsilon_x \tag{6}$$

where

- $K_1 = -3eB\alpha_1T^3/\rho C_p$
- $K_2 = \alpha_2/\alpha_1$,
- v_{xy} = Poisson's ratio in the x - y plane,
- ϵ_x = the strain in the x direction, and

$$\begin{aligned} Q_{11} &= \frac{E_{11}}{1 - v_{12}v_{21}} \\ Q_{22} &= \frac{E_{22}}{1 - v_{12}v_{21}} \\ Q_{12} &= v_{12}Q_{22} \end{aligned} \tag{7}$$

where E_{11} , E_{22} , v_{12} , and v_{21} are the longitudinal and transverse Young's moduli and major and minor Poisson's ratios, respectively, of a unidirectional ply.

Equation 6 shows that if the signal $S = 0$, K_2 can be written as

$$K_2 = \frac{v_{xy}Q_{12} - Q_{11}}{Q_{12} - v_{xy}Q_{22}} \tag{8}$$

It was found for carbon fiber/epoxy laminates [29] that, by varying the orientations of the internal plies from 0 to 25°, the Poisson's ratio ν_{xy} changes from 0.31 to 1.1 and the TSA signals on the surface plies of these laminates change from positive to negative or vice versa. The ν_{xy} under the condition of $S = 0$ can be determined from an experimentally established relationship between S and ν_{xy} . After the value of ν_{xy} is found, K_1 and K_2 can be obtained from Eqs 6 and 8. However, laminates with different subply orientations resulting in a variety of Poisson's ratios must be provided for determining the value of ν_{xy} .

To enable the TSA method for stress analysis of composites, a practical and effective calibration method is presented here. Note that Eq 4 can also be written as

$$S = K_1(\Delta\sigma_1 + K_2 \Delta\sigma_2) \quad (9)$$

where

$$K_1 = -3eB\alpha_1 T^3 / \rho C_p \text{ and}$$

$$K_2 = \alpha_2 / \alpha_1$$

Considering an off-axis unidirectional composite subjected to axial loading as shown in Fig 1, the stresses $\Delta\sigma_1$ and $\Delta\sigma_2$ can be expressed as

$$\Delta\sigma_1 = \Delta\sigma_x \cos^2\theta \quad (10)$$

$$\Delta\sigma_2 = \Delta\sigma_x \sin^2\theta \quad (11)$$

Substituting Eqs 10 and 11 into Eq 9 leads to the expression [30]

$$S = K_1 \Delta\sigma_x (\cos^2\theta + K_2 \sin^2\theta) \quad (12)$$

When the longitudinal axis coincides with the x axis, Eq 12 reduces to

$$S = K_1 \Delta\sigma_1 \quad (13)$$

When the off-axis angle θ (Fig 1) is 90°, Eq 12 has the form

$$S = K_1 K_2 \Delta\sigma_2 \quad (14)$$

The calibration factor K_1 and the material constant K_2 can be determined from Eqs 13 and 14 by testing two specimens.

To verify the method and determine the calibration factor and the material constant for a common material, eleven specimens, 51 mm wide and 1.3 mm thick, cut from a unidirectional glass fiber/epoxy composite were tested. The angle θ ranged from 0 to 90°. The cyclic stresses $\Delta\sigma_x$ applied to these specimens were 33 MPa with a stress ratio of 0.1 at a frequency of 20 Hz. The specimens with 0 and 90° were used to determine the calibration factor and the constant, giving $K_1 = 5.64$ mv/MPa, and $K_2 = 4.62$. Based on the signal S (average values from the scanned area shown in Fig 1), $\Delta\sigma_x$ was calculated from Eq 12, and then $\Delta\sigma_1$ and $\Delta\sigma_2$ were determined from Eqs 10 and 11. The stresses obtained by the TSA method and the prediction (directly substituting the applied cyclic stress $\Delta\sigma_x$ into Eqs 10 and 11) are in good agreement, as shown in Fig 2.

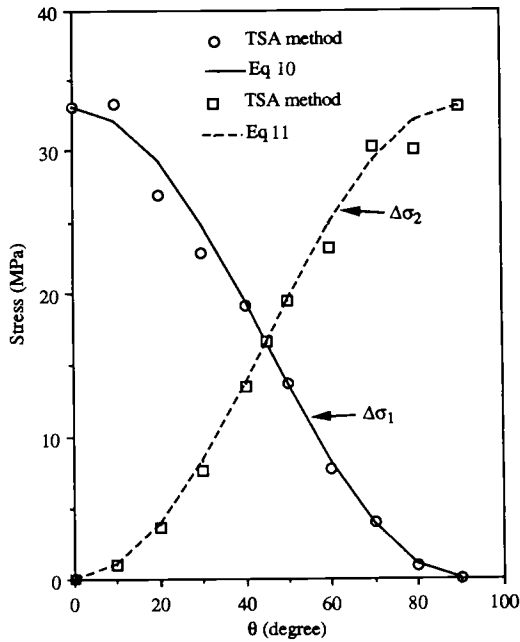


FIG. 2—Comparison of stresses, measurements by the TSA method, and predictions.

ment analysis. Because of the small value of α_1 , the term $\alpha_1 \Delta\sigma_1$ was neglected when the loading direction was coincident with $\Delta\sigma_2$ on the surface ply. Equation 4 was simplified as

$$S = - \left(\frac{3eBT^3}{\rho C_p} \right) \sigma_2 \Delta\sigma_2 \tag{15}$$

The distributions of $\Delta\sigma_2$ were determined from Eq 15. Good agreement between the numerical analysis and the TSA results was obtained on the unnotched ligament in the circular notched specimen. They also superimposed the stress contour lines obtained by the finite element analysis on the thermographic stress patterns and claimed that close agreement between them was found.

In another special case, Zhang and Sandor [31] successfully used the TSA method to estimate the stress concentration factors at the knit yarn/fiber intersections of a unidirectional carbon fiber/epoxy composite.

Stress (Strain) Separation

Wong [32,33] developed a nonadiabatic theory for the thermoelastic effect in laminated composite plates. It has been experimentally and numerically shown that subsurface ply orientations in a composite laminate can significantly influence infrared emissions of the surface ply, depending on loading frequency and ply lay-up. True adiabatic conditions may not be achieved over the usual operational frequency range (5 to 30 Hz). The heat-transferring characteristics of composites under cyclic loading make the TSA method powerful for determining strain components of laminated plates from thermoelastic data.

An experimental and numerical hybrid method has been developed for determining indi-

vidual stress components [34] in an orthotropic tensile plate containing a circular hole from thermoelastically measured data. The smoothing finite element representation (SFER) technique [35] was successfully used for processing the TSA signals from the full scanned field and for eliminating the high-level random noise. The technique also enhances the reliability of the measured boundary data.

Qualitative Evaluation of Damage

The damage mechanism in composites generally consists of the development and coalescence of numerous defects, debonding between fibers and matrix, propagation along the interface and through the matrix, and, finally, failure of the fibers. The TSA method has been used to assess impact damage in composites [36,37], to monitor damage development of unnotched and notched laminates [21,22,30,38–42] and bolted joints [43], and to detect debonding of bonded joints [44].

Impact damage is evidenced by a significant decrease in TSA signals due to lower stress/strain levels than those in the surrounding undamaged material. Full-field thermographic maps corresponding to the damage development of notched composite laminates reveal that decreased TSA signals are caused by load relaxation in the surface ply as a result of fiber failure, matrix crack, and delamination. The combination of the TSA method with X-ray radiography provides valuable insight into the nature of load redistribution in the presence of damage [42]. This method is also quite useful for the evaluation of damage of bolted joints (Figs. 3 through 5).

The geometry of the specimens cut from a glass fiber/epoxy laminate is shown in Fig. 3. The specimen has a width-to-hole-diameter ratio of 4.5, an edge distance-to-hole diameter ratio of 2.7, and a thickness of 3.6 mm. The tests were conducted using a servohydraulic machine. After damaging cyclic loading was applied to the specimen, the infrared equipment was used to scan the damaged regions (Fig. 3) under a nondestructive cyclic stress range of $\Delta\sigma = 10$ MPa with a stress ratio of $R = 0.1$ at a loading frequency of 10 Hz.

The typical failure mode of a bolted joint of $[90/0/90]_s$ laminate is tensile failure. TSA scans using a cyclic stress range of 42 MPa are shown in Fig. 4. The patterns of damage development at the locations of high stresses are illustrated in Fig. 4 (middle) and (bottom). The horizontal light areas (low TSA signals) surrounded by dark areas (high TSA signals) are the critical regions involving two damage mechanisms: transverse matrix cracking in the 90° ply (surface ply) and delamination in the $0^\circ/90^\circ$ interface along transverse cracks. These damaged regions grow toward the free edge of the specimen on the right side as the number of cycles increases. Line plots on Section AA' [Fig. 4 (top)] are illustrated in Fig. 5.

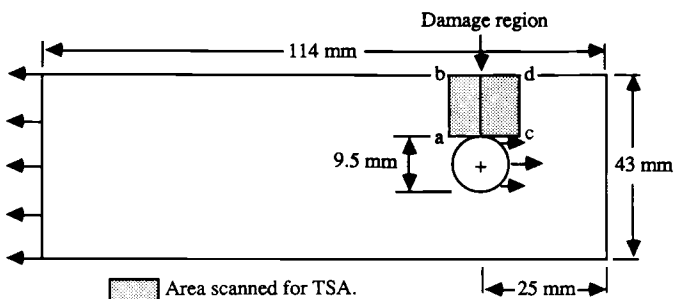


FIG. 3. Geometry of bolted lap-joint specimen, $t = 3.6$ mm.

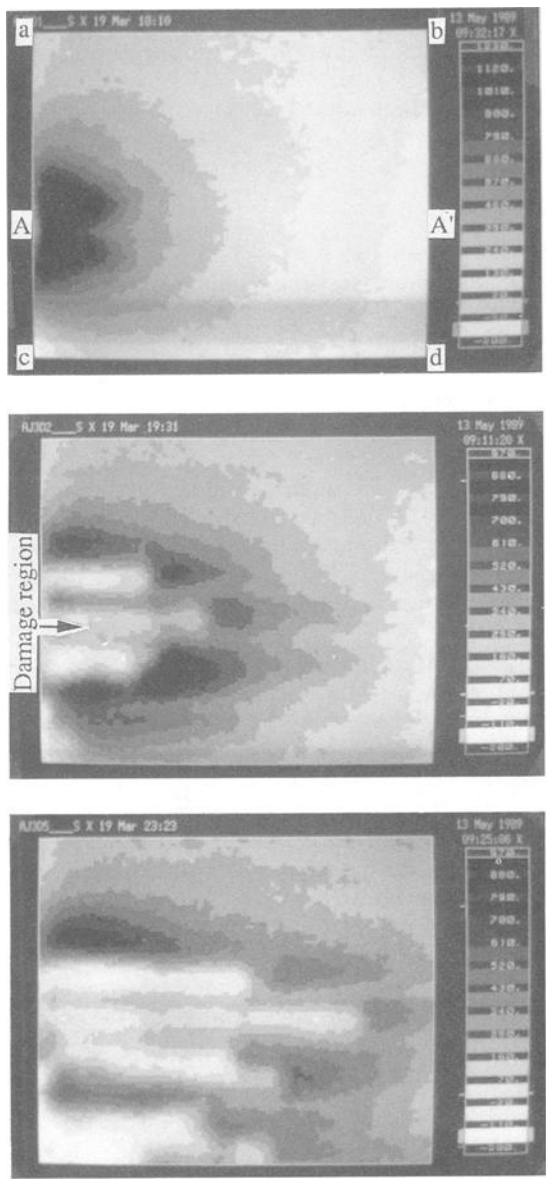


FIG. 4—TSA area scans in a $[90/0/90]_s$ glass fiber/epoxy specimen showing tensile failure mode. Scanned region shown according to Fig. 3; scan area, 16 by 15 mm. (top) TSA area scan of an undamaged specimen. (middle) TSA area scan showing damage development at $N = 830$ cycles. (bottom) TSA area scan showing damage development at $N = 2300$ cycles.

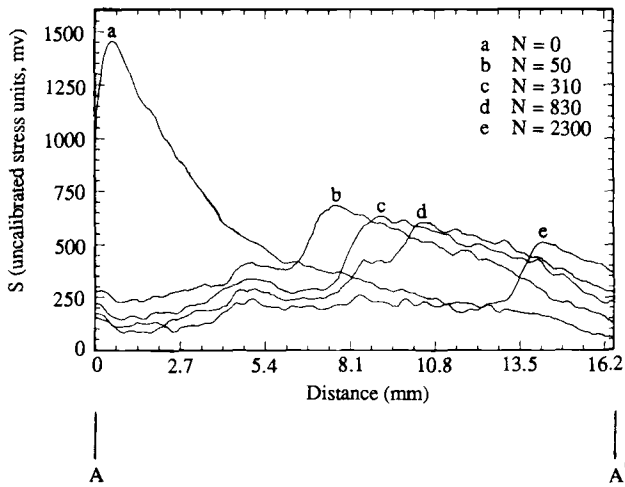


FIG. 5—Line plots on Section AA' according to Fig. 4a.

Quantitative Evaluation of Damage

The TSA method presented in the previous section for damage evaluation is basically the method of uncalibrated stress pattern or temperature change analysis. After damage occurs in these laminates, parameters of damage cannot be determined. To exploit new techniques for the evaluation of damage accumulation, a theory of thermoelasticity for damage in anisotropic materials was developed [45] by using fundamentals of damage mechanics. The generalized temperature change caused by small linear elastic deformation in anisotropic damage can be expressed as

$$\Delta T = \frac{T}{\rho^e C_e} \left(\frac{\partial M_{ij}}{\partial T} C_{jk} \epsilon_k + M_{ij} \frac{\partial C_{jk}}{\partial T} \epsilon_k - M_{ij} C_{jk} \alpha_k \right) \Delta \epsilon_i \quad (16)$$

$i, j, k = 1, 2, \dots, 6$

where

M_{ij} = the effective damage tensor (a function of the damage parameters),

C_{jk} = the stiffness matrix,

α_k = the thermal expansion coefficients,

ρ^e = the effective mass density of the material, $\alpha_k \neq 0$ for $k = 1, 2$, and 3 , and $\alpha_k = 0$ for $k = 4, 5$, and 6 (because the coefficients of thermal expansion affect only extensional strains, not the shear strains), and

C_e = the specific heat capacity for $\epsilon = 0$ [45].

A relationship between the undamaged material's density ρ , effective density ρ^e after damage, and damage parameter D can be expressed in the following form [45]:

$$\frac{1}{\rho^e} = \frac{1}{\rho} \left(\frac{1}{1 - D} \right) \quad (17)$$

Damage is a macroscopic internal variable, and any definition of it can be used if it is a good indicator of the degradation mechanism. It is advantageous that D can be measured through variations of Young's modulus and Poisson's ratios, that is,

$$D = \frac{E - E^e}{E} \tag{18}$$

where

E^e = Young's modulus of the damaged material, called the effective modulus, and
 E = Young's modulus of the undamaged material.

Simplifying Eq 16 by considering a case of uniaxial stress, and substituting it into Eq 3, the signal S can be expressed as

$$S = C_f \left[\frac{\Delta\sigma}{(1 - D)^2} \right] \tag{19}$$

where $C_f = -3eBT^3\alpha/\rho C_p$ is the calibration factor corresponding to the state of undamaged material.

The material used here was a glass fiber/epoxy laminate with $[0/0/90_5]_s$ lay-up. The specimens, cut from the laminate, were 178 mm long and 25.4 mm wide. Aluminum tabs 5.1 mm long were bonded on the ends of the specimens. A cyclic stress range of $\Delta\sigma = 58$ MPa with stress ratio of $R = 0.1$ was applied to produce damage. To avoid inducing extra damage, the TSA signals were taken at a nondestructive cyclic stress range of $\Delta\sigma = 8$ MPa with stress ratio of $R = 0.1$ at a loading frequency of 10 Hz for all full-field TSA scans. The signal used for damage evaluation is an average value from the area scanned for infrared emissions (Fig. 6). With an increase in damage D , Eq 19 gives an increase of signal S . The values of the damage parameter D were obtained directly from Eq 19 as shown in Fig. 7. The normalized mass den-

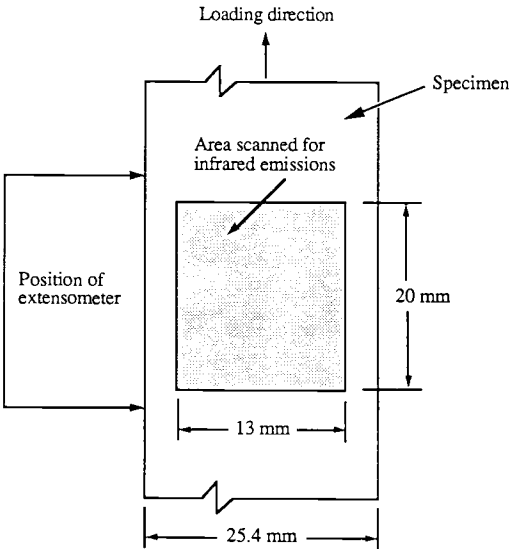


FIG. 6—Position of extensometer and area scanned for TSA.

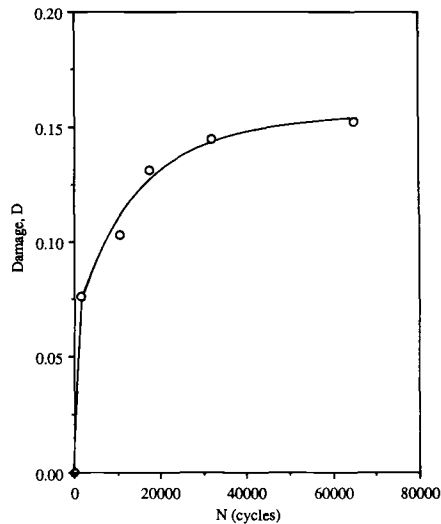


FIG. 7—Damage accumulation of $[0/0/90_s]_s$ laminate measured by the TSA method. Damaging stresses: $\Delta\sigma = 58$ MPa, $R = 0.1$.

sity ρ^e/ρ was computed from Eq 17 and is shown in Fig. 8. The effective Young's moduli obtained by the TSA method (using Eqs 18 and 19) and also measured by an extensometer are presented in Fig. 9. The results from the two different measurements are in good agreement. The theory has also been applied to damage evaluation of glass fiber/epoxy laminates with $[0/90]_s$, $[0/90/0/90/\bar{0}]_s$, and $[0/90_s]_s$ lay-up [45–48]. Note that if the signal S decreases with an increase in damage, which is possible for some materials, the definition of the damage parameter [45,46] must be modified.

Discussion

The interpretation of the TSA signals from composites is more complex than from isotropic materials, as we have discussed. However, the factors that may affect the TSA signals are not always present simultaneously. For example, in glass fiber/epoxy composites, the thermal expansion coefficients do not cause difficulty in stress analysis. If two thermal expansion coef-

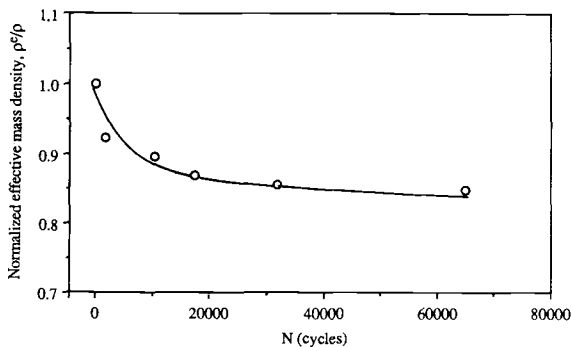


FIG. 8—Normalized effective mass density ρ^e/ρ of $[0/0/90_s]_s$ laminate. Damaging stresses: $\Delta\sigma = 58$

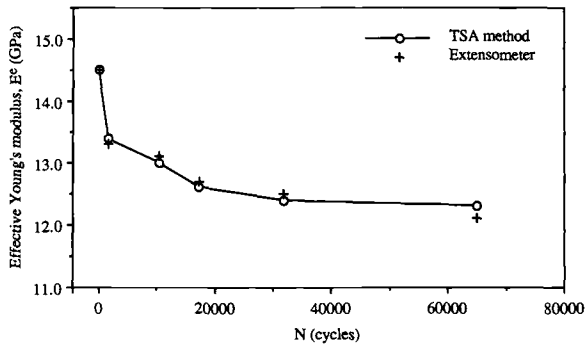


FIG. 9—Comparison of effective moduli of $[0/0/90_s]_3$ laminate measured by two methods. Damaging stresses: $\Delta\sigma = 58 \text{ MPa}$, $R = 0.1$.

ficients and nonuniform fiber densities are involved, the signs of the stress distribution are not easily determined by the TSA method alone, but in some special cases satisfactory results can be obtained [21,31]. The nonadiabatic condition for epoxy-matrix composites extends to much higher frequencies than for isotropic metals. Nevertheless, under a given loading frequency condition, some practical problems can be solved using the TSA data. Because of limited information concerning the effects of absolute temperature increase, surface emissivity variations and surface coating thickness on the TSA signals, further investigation of these subjects is necessary. It is also necessary to understand mean stress effects, although so far the effect of mean stress on the TSA signal appears to be negligibly small in comparison with the effect of cyclic stress range.

Compared with earlier calibration methods, the calibration method presented here using a unidirectional composite is simple and practical. The values of K_1 and K_2 should also be useful in the stress analysis of multiply laminates made of the same lamina, if adiabatic condition can be ensured.

The thermoelasticity theory for damage evaluation and experiments presented here and elsewhere [45–48] have proved the applicability of this method for orthotropic composites. Because of the assumptions made in the derivation of this theory [45,46], it is required that the signal S increase with an increase of damage D . The fibers of the surface ply must be oriented in the loading direction (as in most of the practical applications of composites) to satisfy this requirement; otherwise, the definition of the damage parameter must be modified.

Conclusions and Recommendations

1. The review shows that significant progress has been made in using the TSA method for composites, and it clarifies the need for improved interpretation of TSA signals.
2. The calibration method using a unidirectional composite involving a calibration factor and a material constant is practical and effective, and the stress analysis using the TSA method on off-axis glass/epoxy composites has shown good results. The calibration factor and the material constant can also be used for the stress analysis of multiply composites, but it needs further verification.
3. The TSA method has been used successfully in the qualitative evaluation of damage in a variety of composites with different geometries and structures.
4. Initial progress in stress (strain) separation techniques for laminated composites has been achieved.

- 5 The TSA method based on the thermoelasticity theory for anisotropic damage can be used to measure damage quantitatively in some fiber-reinforced composites, but the applicability of this method to other composite laminates needs further verification
- 6 Recommended future research topics are combination of numerical techniques and the TSA method for full-field stress (strain) analysis, further development of stress (strain) separation methods for practical applications and more complex geometries; and quantitative evaluation of damage and detection of subsurface defects at elevated temperatures in composites

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Stress Field Sensitivity of a Composite Patch Repair as a Result of Varying Patch Thickness

REFERENCE: Siener, M. P., "Stress Field Sensitivity of a Composite Patch Repair as a Result of Varying Patch Thickness," *Composite Materials: Testing and Design (Tenth Volume)*, ASTM STP 1120, Glenn C. Grimes, Ed., American Society for Testing and Materials, Philadelphia, 1992, pp. 444-464

ABSTRACT: The desired method for repair of composite airframe structures is by joining with an adhesive(-like) material. The available choices of structurally efficient joint configurations are limited. Most of the commonly used joint configurations would result in repairs that would be either impractical, because of the long adherend overlap lengths required to sustain the design load, or unsafe, because a shortened lap length would limit the load-carrying capability of the joint. Single lap joint configurations based on a variety of patch material thicknesses and stiffnesses are investigated for elastic response and strength capability. The section stiffness of the patch material is kept constant in the direction of the first principal stress, as the number of plies is decreased. The average modulus of the patch adherend is increased by reorienting the plies in the laminate. This is done to demonstrate the possibility of increasing the average joint failure shear stress and the accompanying load transfer efficiency by increasing the stiffness of the adherends. Stress analyses of these configurations are carried out and compared to the results from testing the described configurations.

Experimental data show that it may be necessary to incorporate the nonlinear adhesive behavior into the model to account for the extensional response of the adherends. Also, it may be necessary to model each composite ply to predict the extent of incompatibility of the adhesive bondline and the adjacent composite material. The data also demonstrate that the failure of a composite joint depends on the strength of bondline-contacting interfacial plies, as well as the usual factors considered when estimating joint strength.

KEYWORDS: composite materials, composite repair, composite patches, joints, adhesives, bondlines, scarf lap joints, stepped lap joints, single lap joints, finite element models, static tests, strength, laminates, adherends

Nomenclature

C	Constant of proportionality
CL	Centerline
D, E, F, G	Weight factor between 0 and 1.0 that dictates extent of strength loss
E_x	Modulus of a ply in the longitudinal (fiber) direction
E_y	Modulus of a ply in the transverse (matrix) direction
E	Young's modulus
F_{xtu}	Ultimate tensile strength of a ply in the longitudinal direction
F_{ytu}	Ultimate tensile strength of a ply in the transverse direction
F_{sysu}	Ultimate in-plane shear strength of a ply
F_{lu}	Ultimate tensile strength

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F_{su}	Ultimate shear strength
FE	Finite element
FEA	Finite element analysis
FEM	Finite element model
G_{xy}	In-plane shear modulus of a ply
G	Shear modulus
k	Numerator of exponent that dictates severity of the reduction of the rate of load increase with lap length
L_{bondline}	Length of the bondline
N_1	First principal stress resultant
N_{failA}	Stress resultant that causes adherend laminate failure, calculated using Tsai-Wu criteria
N_{failB}	Stress resultant that causes failure in bondline, load extrapolated from finite element results, which corresponds to the peak shear stress in the bondline equaling the ultimate shear strength of the adhesive
N_{failC}	$F_{su} \times L_{\text{bondline}}$
N_{failD}	Shear stress resultant at failure of specimen recorded from experiment
N_{iso}	Failure stress resultant of a fully scarfed bonded joint with isotropic adherends
N_{comp}	Failure stress resultant of a partially scarfed joint with composite adherends
n	Number of plies in the laminate
n_{90}	Number of 90° plies in the laminate
n_{45}	Number of 45° plies in the laminate
n_0	Number of 0° plies in the laminate
t_{ply}	Thickness of one-ply composite material
t_{step}	Thickness of a step in truncated scarf joint
t_{lam}	Thickness of laminate
ϵ_y	Extensional failure strain of a ply in the transverse direction
x	Position along overlap
δ_1	Displacement of Adherend 1
δ_2	Displacement of Adherend 2
ϵ_{fail}	Extensional failure strain of the adhesive
γ_{bondline}	Shear strain within the bondline
η	Cured thickness of bondline
ν_{xy}	Major Poisson's ratio of a ply
σ_1	First principal stress
σ_2	Second principal stress
τ	Shear stress

Introduction

The use of fibrous composite materials for aircraft and other structural applications is increasing because of the need to save weight and reduce corrosion. This increased use requires that repair methods become established so that lifetime costs for such structures are reduced and are less than or equal to those for metallic structures. Composite aircraft structures are repaired in the following way. The cracked or delaminated portion of the structure is cut away. New material is installed providing the continuity and strength needed for the aircraft structure to sustain again the load for which it was designed. The installation of this new material causes the formation of a joint, or in the case of a patch repair, two joints, which are used to transmit the load from the parent aircraft structure to the patch material and then to the parent structure again.

Many types of joints have been devised for the joining and repair of structures. Both bonded and bolted joints are used in composite structural repairs. Simple lap, stepped lap, and scarf lap joints define the possible geometries of joints formed as a consequence of replacing the damaged material in the structure with undamaged material. Double lap, double stepped lap, and double scarf lap joints, while specified in new aircraft design when additional strength is needed, are, in general, not used for repair because of manufacturing and fit-up difficulties. Bolted joints lead to stress-raising effects during loading and give rise to delamination around bolt holes during installation. Bonded-bolted joints are inefficient because the bonded joint is much stiffer than the bolted joint and failure of the adhesive bondline layer is likely to occur before the bolts are able to carry their share of the load. Consequently, adhesively bonded joints are preferred over bolted joints for a composite repair.

The optimum bonded repair patch incorporates a beveled bonding surface onto the patch and parent structure to gain the structural advantage of a scarf lap joint. The scarf lap joint is desirable because it minimizes the peak shear and peel stresses at the end of the bondline and hence increases the efficiency and the accompanying load-carrying capability of the joint. To return a significant percentage of the laminate strength through a bonded patch repair, the overlap length that provides the bonding surface must be inordinately long. The bonded repair designer frequently must decide whether to size the patch with a large overlap area, so that significant strength returns may be attained, or to size the patch so the overlap length is small, allowing repair area surface preparation and installation problems to be minimized (the resulting repaired structure has much less load-carrying capability than the original structure).

This paper investigates the possibility of increasing the efficiency of a composite scarf joint and, hence, composite repair. The investigation is carried out by analyzing and testing five different joint configurations for elastic response and strength capability. Correspondingly, five finite element models (FEM) are used for analysis of the composite lap joint specimens. The analyses are verified by the testing of the lap joints and then a comparison is done. The validity of the analyses is discussed, and recommendations are made on how to analyze the strength of composite scarf lap joints.

Specimen Design

When a composite patch repair is viewed from the side, two identical lap joints are evident (Fig. 1). The scarf ratio and thickness of the patch define the length of the joint in the repair. Joint efficiency should be maximized so that the length may be minimized, thereby reducing mechanic rework time and associated costs and aircraft down time, while preserving the structural integrity of the joint. The scarf ratio for a composite scarf repair is usually from 20:1 to 50:1. That is, the length of the lap joint recommended in a repair where the scarf ratio is 50:1 is 50 times the thickness of the laminate. As an example, a 2-in. (5-cm) through-hole in a 32-ply laminate, using a scarf ratio of 50:1 would require the impractical application of an 18.6-in. (47.2-cm) diameter circular composite repair patch.

The overlap length in the repair can be shortened by truncating the joint. The length of the adherend is scarfed beginning somewhat further along the length of the adherend so that the end of the adherend does not form a knife edge. Rather, the end of the adherend has some finite thickness.

Lap joint specimens were analyzed and tested to investigate the structural response of a composite repair. The specimens analyzed and tested reflect the geometry that will be used for composite repair patches in the future (Fig. 2). Except for Specimen 1, all the lap joints tested have a truncated scarf joint which, while reducing overlap length, introduces a significant stress raiser, which may contribute to premature joint failure. The following is a description of each specimen. Specimen 1 is the baseline specimen because it incorporates a fully scarfed

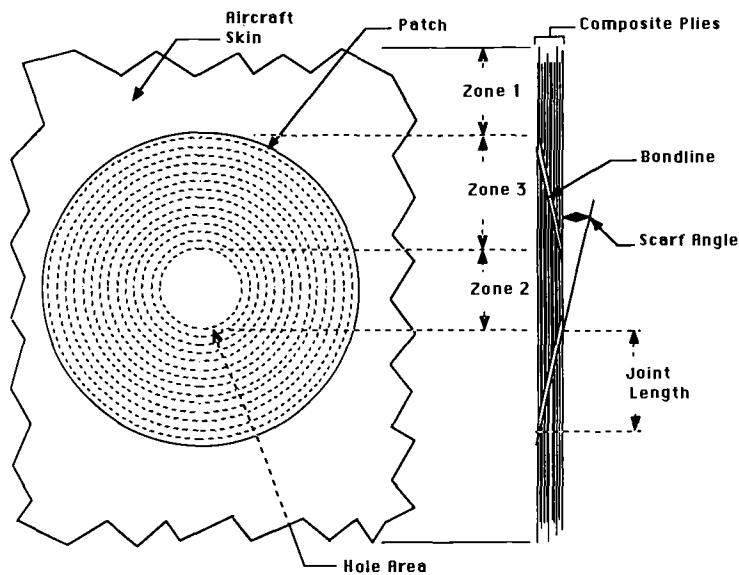
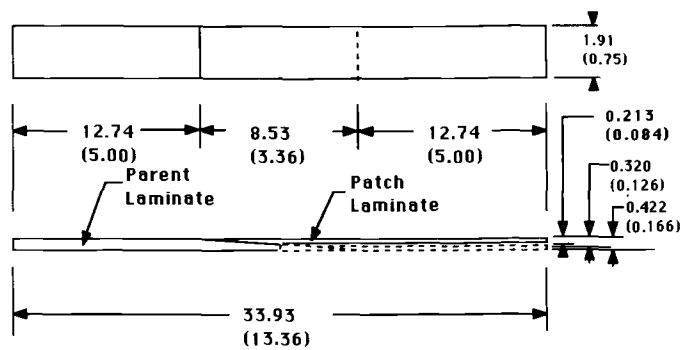


FIG. 1—Composite scarf patch repair.



Specimen	Laminate	E_1 Mpa (psi)	t cm. (in.)	Section Stiffness($E_1 \times t$) MN/M (lb./in.)
1	[0/+45/90]4s	5.02E4 (7.28E6)	0.422 (0.166)	211.8 (1.21E6)
2	[0/+45/90]3s	5.02E4 (7.28E6)	0.320 (0.125)	160.6 (0.91E6)
3	[0/+45/90]2s	5.02E4 (7.28E6)	0.213 (0.083)	106.9 (0.60E6)
4	[0,3/90/[0/+45/90]2]s	6.81E4 (9.88E6)	0.320 (0.125)	217.9 (1.23E6)
5	[0,6/90,2]s	10.2E4 (14.8E6)	0.213 (0.083)	217.3 (1.23E6)

FIG. 2—Test matrix showing the five specimens analyzed and the physical properties of each. All dimensions on drawing are in centimetres (inches).

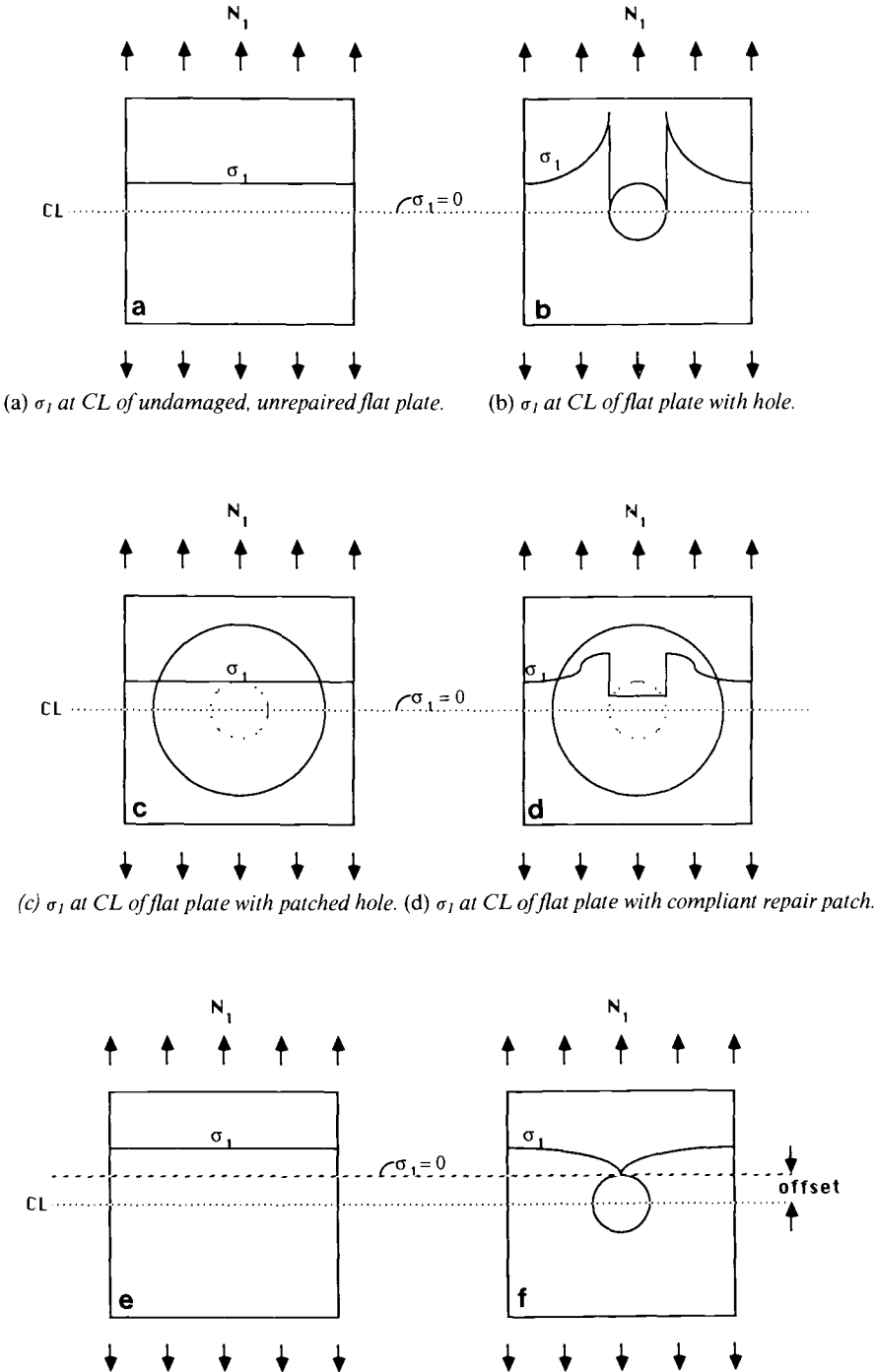
joint and is expected to be the strongest. Both patch and parent adherends have 32 plies. In the other four specimens, the ply count is varied in the patch adherend, but the parent adherend remains 32 plies. Specimen 2 has the same parent laminate, but reduces the ply count and hence the thickness and lap length of the patch laminate by 25%, while retaining the quasi-isotropic laminate. Specimen 3 further reduces the ply count in the patch adherend by 25% of the original 32 plies. This corresponds to an additional 25% less thickness and lap length than Specimen 2. Specimen 4 has the same overall geometry (not ply angles) as Specimen 2 but retains the same overall patch laminate section stiffness in the 0° laminate direction. This is done by orienting plies so that there are a greater percentage of 0° plies in the patch laminate than in the parent laminate. Hence, there is a stiffness loss in the 90° laminate direction in those patch laminates where stiffness in the 0° laminate direction has been increased. Specimen 5 has the same overall geometry as Specimen 3 but retains the same overall patch section stiffness as the parent adherend.

The philosophy behind specimen design was driven by a rule-of-thumb for patch design. For an efficient repair design, the patch stiffness should match the parent material stiffness. This rule is echoed by Hart-Smith [1–4] who states that joint efficiency is maximized when the adherends are of equal stiffness. The idea is further illustrated in Fig. 3. Consider the stressed condition of a uniformly loaded isotropic (or quasi-isotropic) flat plate, which might be representative of a portion of an aircraft skin before damage and patch application. The normal stress, σ_1 , remains constant or has a slight gradient along the width of the plate (see Fig. 3a). (This slight gradient may exist due to the changing bending moment along the length of the structure.) The stress field of the plate with an open hole is known, it is the condition that exists after the damaged portion of the plate has been removed (see Fig. 3b). The stressed condition of a patch-repaired composite plate should not differ from the stress field of an undamaged plate under the same load. This is possible if the modulus and thickness of the replacement material match the material that has been removed (see Fig. 3c). If, however, the patch is more compliant than the laminate that was removed, σ_1 at the edges of the hole (see Fig. 3d) is raised.

In fact, Fig. 3d illustrates that σ_1 approaches the limiting function along the width of the plate equivalent to the stress field of a plate with an open hole as the stiffness of the patch approaches zero. So, in this repair case, the continuity of the laminate has been restored by application of the patch, but the stress field in the plate has not been brought back to its original state—one in which there are no high stress gradients that could cause premature rupture of the laminate. Next, consider the case of a patch that is stiffer than the surrounding parent material. In this case, the critical location of the first principal stress, σ_1 , is at an offset from the centerline of the plate. Because the plate does not change cross section, σ_1 at the offset is the same as that at the centerline of an undamaged plate (see Fig. 3e). The σ_1 stress field at this location in a plate with an open hole is a minimum at the center of the plate, because the lack of material provides no stress transfer (see Fig. 3f). The desired stress field at this location in the repaired plate is one where the σ_1 profile matches the undamaged plate (see Fig. 3g). When the stiffness of a repair patch is greater than the surrounding laminate, the stresses carried by the laminate to either side of the patch are lowered, forcing the load path through the patch and the laminates in front of and behind the patch (see Fig. 3h). Additionally, because of the increase in patch stiffness, the bondline that transfers the load into the patch carries higher shear stresses. This shear stress escalation in the bondline is definitely an undesirable result of patch application because the bondline may be considered the “weakest link” of all elements in an aircraft patch repair.

With the preceding discussion in mind, further consideration of the test specimens is needed. Specimen 1 represents a patch installed to replace the plies that have been damaged and removed. Because, in this case, the plies that were removed have been replaced, the stiff-

FIG. 3— σ_1 at CL or at offset in plates at various stages of repair.



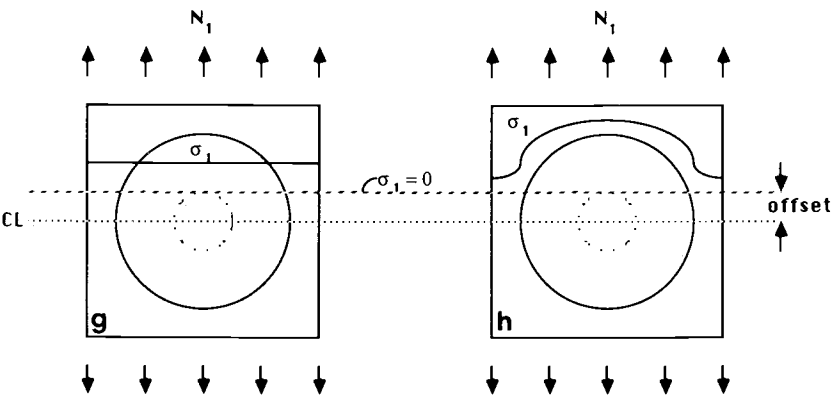


FIG. 3—(g) σ_1 at offset of flat plate with patched hole. (h) σ_1 at offset of flat plate with stiff repair patch.

ness of the repair matches the surrounding laminate and, without considering the strength of the bondline, provides the laminate stiffness and strength needed. Specimens 2 and 3, however, do not provide the stiffness required, so they must be eliminated as candidates for repair patches. They are retained in the test spectrum because their overall geometry matches Specimens 4 and 5, and it is advantageous to compare the strength efficiencies of the specimens with similar overall geometries. This comparison is important because it was anticipated that additional strength efficiency may be gained by increasing the stiffness of the adherends, thus causing a decrease in the maximum stresses at the end of the bondline that results from the parabolic shear stress behavior.

Analytical and Experimental Methods

The method for analyzing the five lap joints specified in this paper was through finite element analysis (FEA). Five FEM were developed (Fig. 4), one for each specimen in the test spectrum. MSC/NASTRAN [5] was used as the solver; PATRAN [6] was used to construct and postprocess the models. The modeling philosophy was consistent for all specimens and consisted of the following: Linear statics was used as the solution technique. Material nonli-

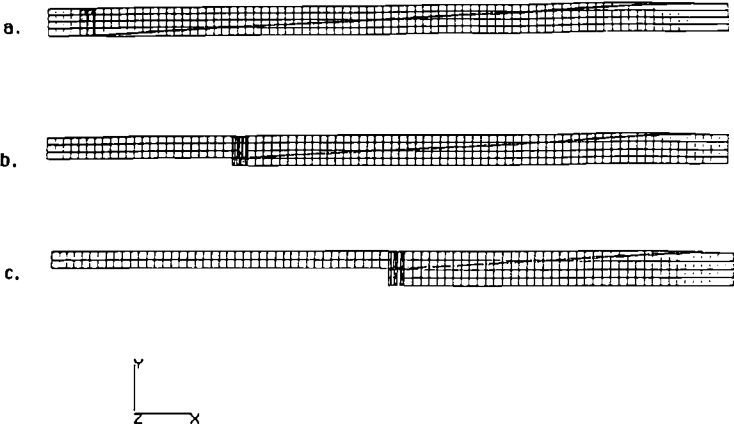
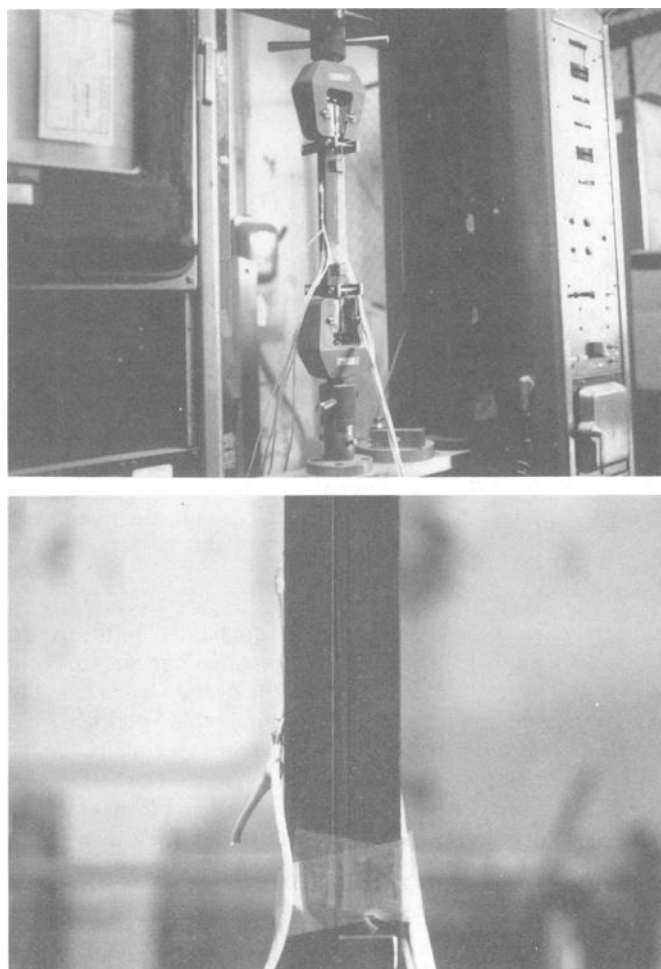


FIG. 4—Finite element models of the specimens. (a) Mesh used for Specimen 1. (b) Mesh used for Specimens 2 and 3. (c) Mesh used for Specimens 4 and 5.

nearity was not taken into account. In addition, interply adhesive (matrix) was not modeled; neither were the individual plies. The analysis performed was isotropic. The Young's modulus assigned to the adherend elements was the average modulus calculated from a laminated plate analysis. Two-dimensional elements, mostly quadrilaterals, were used, and plane strain boundary conditions were assumed. The mesh density was increased at the end of the joint where stress gradients are expected to be the highest. Also, the bottom nodes on each model were constrained in the y -direction, which simulates the support that a joint might receive from a honeycomb core, a rib, or the material from the parent structure surrounding the patch. All the degrees of freedom along the left edge of the model were fixed, simulating a clamped boundary condition. The load on each model was imposed by a required displacement to the right by the nodes on the right edge of the model yielding an average of 3500 μ -strain along the length of the specimens.

Each specimen was tested using an Instron tension testing machine (Fig. 5). A support fixture (Fig. 6) was designed and used to prevent bending in the specimen, thereby simulating the



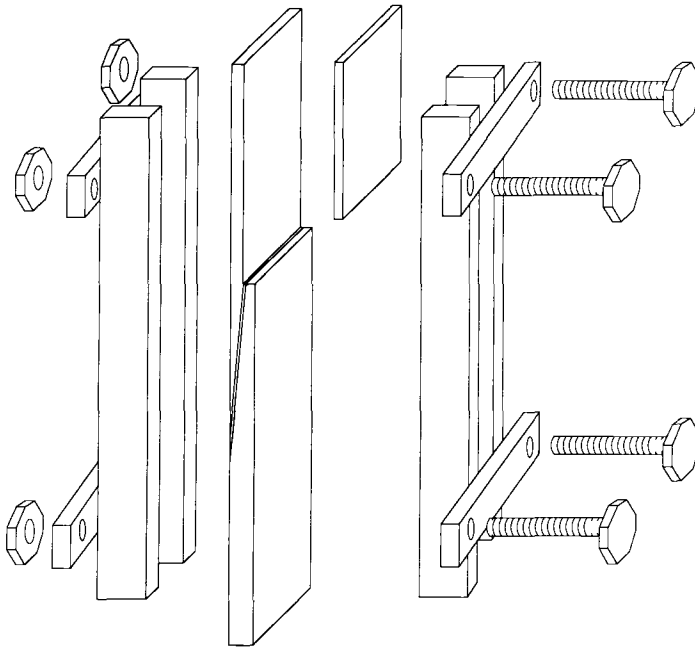


FIG. 6—Exploded view of specimen and fixture.

boundary conditions imposed on the FEM. Each side of the fixture consisted of two vertical beams, which would allow the passage of strain gage wires and resist the small bending moment, and two horizontal beams, which would allow bolts to pull the two sides of the fixture together onto the specimen. A spacer was used for Specimens 2 through 5 to fill the volume in the grips that was vacated by the missing plies. The bolts in the fixture were hand-tightened to allow for light contact of the fixture to the specimen before loading. This fixture ensured that a moment was not imposed on the joint from the clamping forces of the grips and the misalignment of the centerlines of the adherends. Strain gages were positioned to measure strain along the length of the bondline. Additional testing specifics are presented in Ref 7.

Results and Discussion

The adhesive bondline shear stress FEA results are plotted for the five specimens in Figs. 7 through 9. Upon inspection of the FEA results, Specimen 1 appears to be the most structurally efficient (see Fig. 7). This was expected because the overlap length was the largest of the five specimens. There is the additional structural advantage that Specimen 1 is a fully scarfed joint and does not incorporate the truncation of the other four specimens. Consequently, the results for Specimen 1 illustrate the advantage of a scarf lap joint. The high gradient parabolic shear stress behavior does not occur, resulting in a shear stress profile that is approximately constant along the entire length of the joint. (Richards [3,8] identified less than 4% stress concentrations at the ends of some scarf joints.) The remaining four specimens exhibit large gradient shear stress behavior.

Figure 8 illustrates the displacement of Specimen 2 under load. The deformations shown are the calculated values from the FEA imposed on the existing computer image of the model. These deformations were multiplied by an amplification factor so that the relative behavior

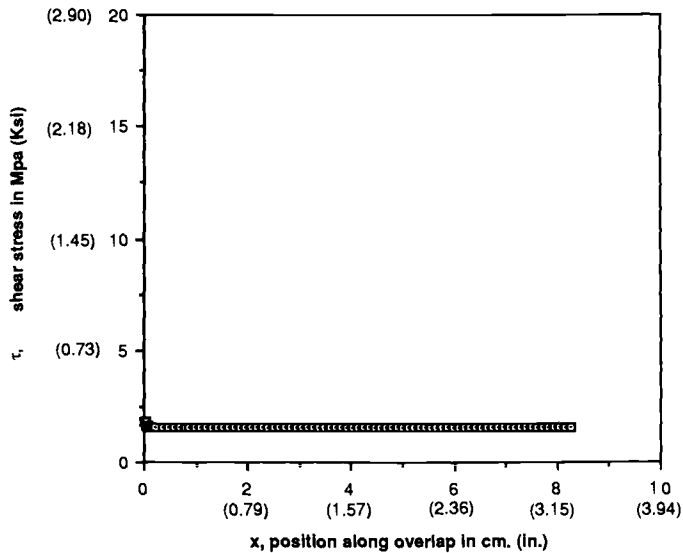


FIG. 7—Bondline shear stress, Specimen 1.

could be more readily seen. Although this plot is for Specimen 2, the deformation behavior shown was similar in the other 5 specimens (that is, small deformations in those elements comprising the adherends and large shear deformations in the elements that make up the bondline). The decreasing shear deformation in the bondline elements with increasing distance from the end of the bondline is consistent with modern bonded joint theory [9,10]. This contrasts with the gradual taper and low stress gradient resident in the scarf joint of Specimen 1.

Figure 9 further magnifies the area of interest in Specimen 2 to illustrate the shear stress field. Consistent with the high bondline deformation gradient shown in Fig. 8, Fig. 9 shows the accompanying high bondline stress gradient. The shear stress field shown in Fig. 9 is similar to those that could be presented for Specimens 3 through 5. The behavior is more fully rep-

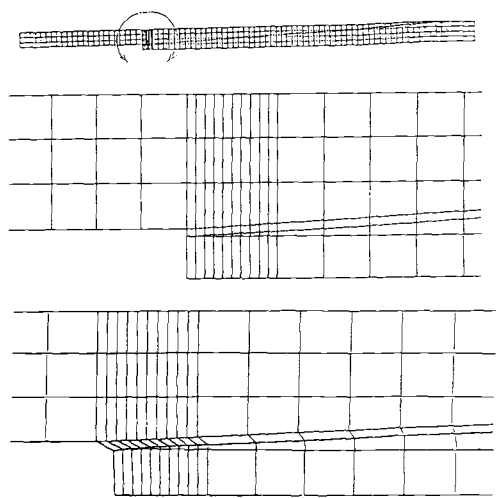


FIG. 8—Undeformed and Deformed plot of critical area of the joint.

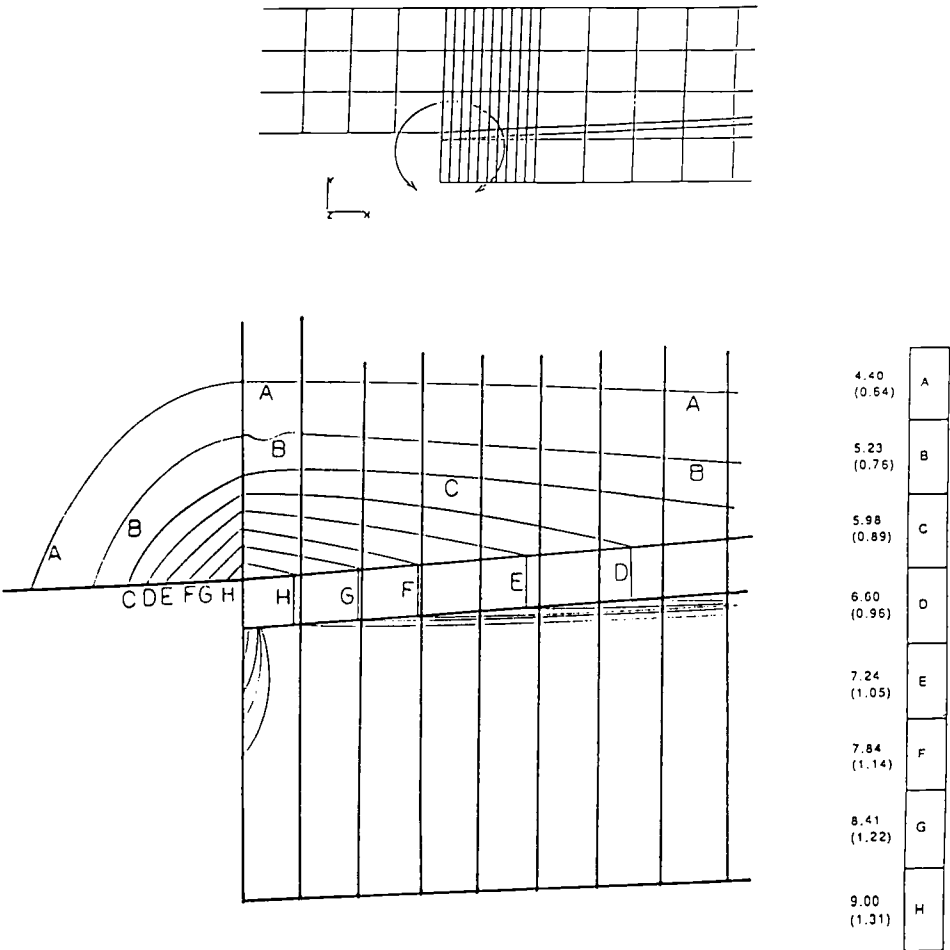


FIG. 9—Shear stress in MPa (ksi) in the critical area of the joint at a load of 2.63 kN/cm (1500 lb/in.).

resented, however, by graphs of the type shown in Fig. 10. In these plots, the magnitude of both the average and the high shear stress at the end of the bondline may be compared among the specimens. These plots show that shear stress is a maximum at the truncated adherend end and a minimum at the middle, which complies with the theories of Volkersen [9] and Goland and Reissner [10]. Because each Joint 2 through 4 had one end that incorporated a tapered adherend and one end that was truncated, the resulting shear stress profile was not parabolic and showed large stress gradient behavior at only one end of the joint. The overall dimensions of Specimens 2 and 4 are the same. So, it is worthwhile to compare their results so that the advantage from substitution of a stiffer patch adherend can be seen. The significant structural advantage is apparent in Specimen 4 because the peak stress of Specimen 2 is larger than that of Specimen 4. This was expected because the elements comprising the patch adherend in Specimen 4 have been assigned a higher modulus than those elements that comprise the adherend in Specimen 2. This larger adherend stiffness for Specimen 4 allowed more even load

a. specimen 2.

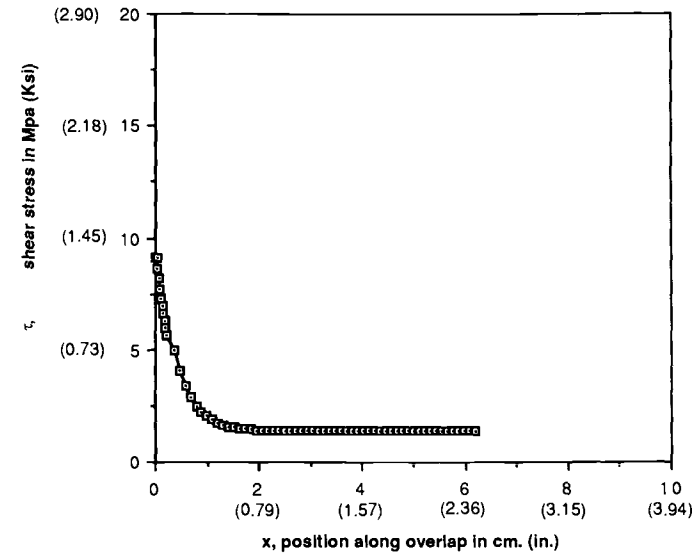


FIG. 10—Finite element bondline shear stress results for Specimens 2 through 5. $N_1 = 2.63 \text{ KN/cm}$ (1500 lb/in.). (a) Specimen 2.

b. Specimen 3.

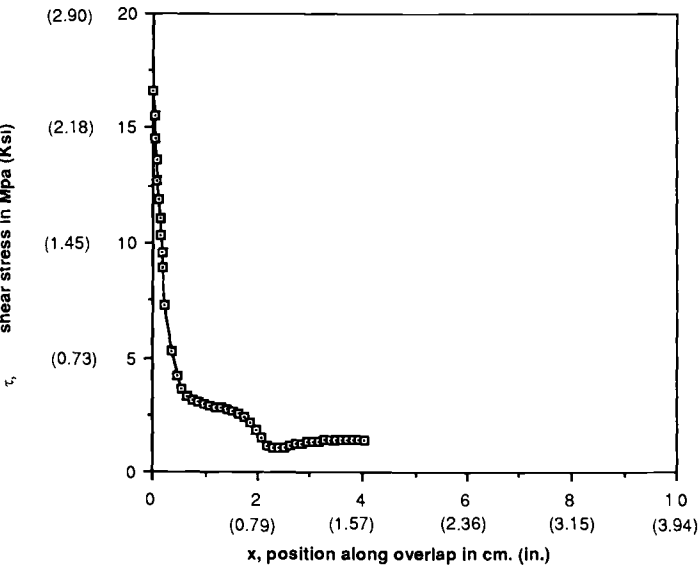


FIG. 10—(b) Specimen 3.

c. Specimen 4.

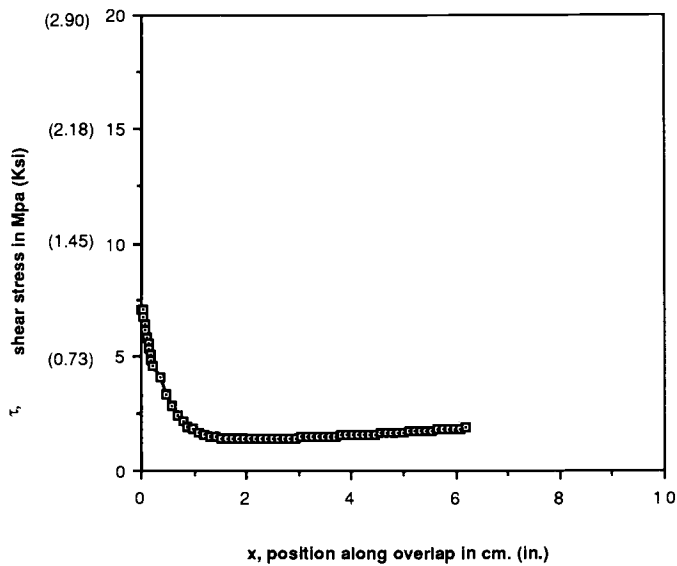


FIG. 10—(c) Specimen 4.

d. Specimen 5.

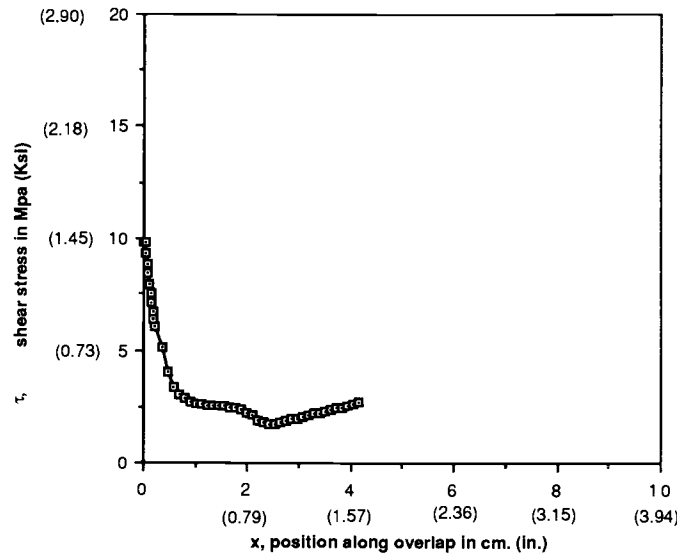


FIG. 10—(d) Specimen 5.

transfer, resulting in higher joint efficiency. As expected, the largest bondline shear stresses occurred in Specimens 3 and 5. This was anticipated because the bondline length is shortest of all the specimens, ensuring a higher average shear stress than the other specimens. In addition, the stress-raising step has the greatest thickness in these specimens, forcing the differential extensional strains in the adherends to increase, resulting in a higher bondline shearing strain. Between these two specimens, a significant structural advantage is seen in Specimen 5, which was expected, because the elements comprising the patch adherend in Specimen 5 have been assigned a higher modulus than those elements that comprise the adherend in Specimen 3.

The original intention for the attachment of strain gages along the overlap length of the bonded joint was to generate experimentally derived bondline shear strain profiles. In Ref 1, the method used for doing this is incorrect. The correct method in using the surface measured extensional strains to calculate bondline shear strains resulted in bondline shear strains that were much higher than expected. Because the strain gages are located at the outer surfaces of the adherends, the shear strain calculated from the measured extensional strains represents the through-the-thickness shear strain of the joined composite laminates. This shearing strain is appreciable due to the relatively compliant composite laminate. This would not be anticipated to be a problem with isotropic adherends of sufficiently high stiffness. Although the strain gages were not useful in calculating bondline shear strains, they were useful in detecting the onset of damage. Typical strain readings would become significantly nonlinear somewhere between 50 and 100% of the failure load. These strain profiles are not presented for the sake of brevity.

The strength results from load testing the five specimens are shown in Fig. 11. As can be seen, Specimen 1 failed at 10.3 KN/cm (5880 lb/in.). Specimen 2, with one fourth less lap length and a stress-raising step built into the overlap, failed at 5.98 KN/cm (3416 lb/in.). Specimen 3, with one half less lap length than Specimen 1 and a larger step stress raiser than Specimen 2, failed at 4.46 KN/cm (2547 lb/in.). Specimen 4, with the same gross geometry as Specimen 2, surprisingly had the highest strength of 12.03 KN/cm (6867 lb/in.). Specimen 5 exhibited the weakest joint strength of 3.97 KN/cm (2269 lb/in.).

In general, all five specimens failed in a similar manner: A crack initiated somewhere in or very close to the bondline and propagated in both directions along the bondline until separation of the patch and parent adherends took place. The path of crack propagation for all specimens, however, did not remain between the bonding surfaces of the adherends for the entire

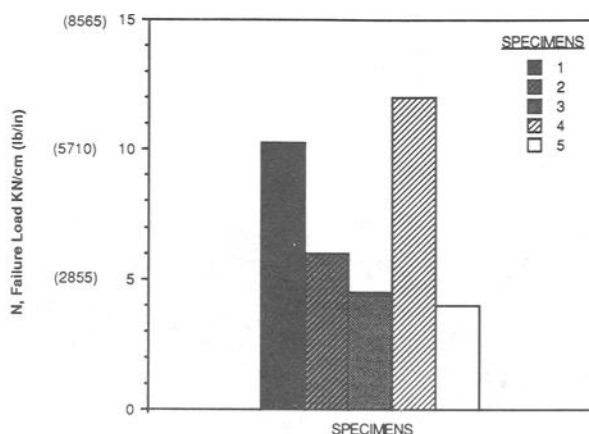


FIG. 11—Failure load results for specimens.

length of the joint. And, the mode of fracture was not either a simple adhesive or cohesive failure within the bondline [11]. During failure of the specimens, the path of crack propagation developed in the patch adherend. In fact, in all cases, the crack followed the bondline in a steplike manner, propagating along the adhesive layer, causing separation of adhesive layer and patch plies, then turning at 90° to propagate through the thickness of several patch plies, then turning again a -90° (negative to the first 90°) to again follow the bondline, only to turn again and move through the patch plies back to the adhesive layer surface and continue propagating along the patch/adhesive interface. Figure 12 shows the typical crack path in an illustrated close-up of the bondline. In this figure and without exception, none of the failed specimens showed the crack propagating through 0° angle plies (that is, no portions of 0° plies were found broken from the patch adherend and part of the fracture surface of the parent adherend). In the case of the 45° plies, some of these plies were removed from the surface of the patch adherend and some were not (although Fig. 12 depicts the crack propagating through all of the 45° plies that were encountered). All exposed 90° plies were torn from the patch adherend surface.

Also, in all cases, the fracture surface reveals that the adhesive remained adhered to the parent adherend, except for a small piece of adhesive adhered to the patch adherend in Specimen 5. In all cases, parts of plies from the patch adherend were torn off and remained affixed to the parent adherend and became a part of the fracture surface. Of the torn-off plies that were examined, all were off axis (nonzero degree) plies. The location of crack initiation could not be determined.

The strains and loads measured while testing each specimen can be accepted with a high degree of confidence because this behavior is highly repeatable and not prone to yield a large scatter in data. Therefore, a high degree of variability would not be expected from strain gage readings if more specimens of one type were tested. This supposition is supported by the fact that the measured far field strains of the parent and patch were within 7% of the calculated values (Table 1). The strength characteristics of the composite lap joint specimens, as opposed to the elastic characteristics, on the other hand, must be considered to be highly variable, and prone to the general high data scatter characteristics associated with strength testing composite laminates. With this consideration in mind, the strength data derived from load testing the five different specimens are not supported statistically in this paper. The values shown in Fig. 11 are a result of testing one specimen of each type. Therefore, the strength recorded by any one specimen must be regarded with suspicion and used in conjunction with the strengths recorded by the other specimens. The strength data of all five specimens as a whole are presented to

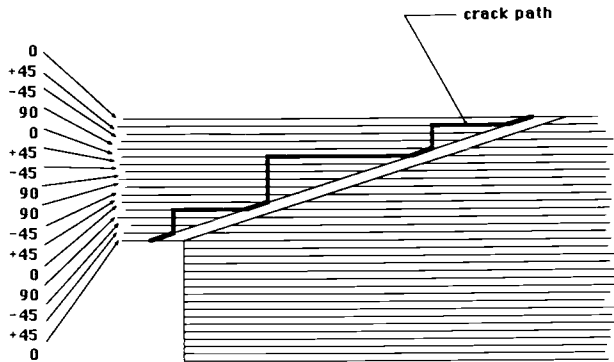


FIG. 12 Characteristic crack path of tested specimens. Specimen 3 shown.

TABLE 1—*Calculated and measured strains in non-overlap section of specimens ($N_1 = 3.50 \text{ KN/cm (2000 lb/in)}$)*

Specimen	Parent Adherend		Patch Adherend	
	$\epsilon_{\text{calculated}}$	$\epsilon_{\text{measured}}$	$\epsilon_{\text{calculated}}$	$\epsilon_{\text{measured}}$
1	0 001 655	0 001 659	0 001 655	NA
2	0 001 655	0 001 608	0 002 061	0 002 079
3	0 001 655	0 001 656	0 003 103	0 003 310
4	0 001 655	0 001 656	0 001 518	0 001 620
5	0 001 655	0 001 656	0 001 527	0 001 629

make some generalizations about the trends of bonded composite lap joints and to make some comparisons of the tested strength and the calculated strength

The relative strengths of the tested specimens generally followed the expected trend. As the overlap length was increased, the joint became stronger (that is, Specimen 1 was stronger than Specimen 2, which was stronger than Specimen 3). And as was expected, the strength of Specimen 5 was comparable to the strength recorded by Specimen 3. The exception to this trend was the strength recorded by Specimen 4. Specimen 4, although resembling Specimen 2 in gross geometry, was much stronger than Specimen 2. In fact, despite having a lap length one fourth less than Specimen 1, and incorporating the step in specimen thickness, Specimen 4 recorded a failure load 16.8% larger than the failure load for Specimen 1. This larger failure load was unexpected because Specimen 1 was thought to be the strongest specimen because of the additional lap length and the lack of a stress raiser. Further strength testing is needed to determine what effects ply angle percentages and ply sequences have on joint strength.

The tested joint strengths of the specimens and several calculated strengths are shown in Table 2. The calculated strengths are based on laminate and bonded lap joint strength calcu-

TABLE 2—*Calculated and actual failure loads of specimens. Load is KN/cm (lb/in)*

Specimen	Calculated			Measured	
	N_{failA}^a	N_{failB}	N_{failC}	N_{failD}	
	FPF	LPF			
1	9.52 (5439)	18.84 (10 756)	45.73 (26 115)	52.96 (30 240)	10.30 (5880)
2	7.17 (4095)	14.18 (8099)	7.76 (4431)	39.72 (22 680)	5.98 (3416)
3	4.76 (2719)	9.42 (5377)	4.27 (2439)	26.48 (15 120)	4.46 (2547)
4	9.52 (5439)	18.84 (10 756)	10.00 (5713)	39.72 (22 680)	12.03 (6867)
5	9.52 (5439)	18.84 (10 756)	12.29 (4128)	26.48 (15 120)	3.97 (2269)

^a Where

N_{failA} = laminate failure load using Tsai-Wu [13] criteria,

N_{failB} = bondline failure based on extrapolated data from the FEM. That load that causes shear stress in any part of the bondline to exceed F_{su} .

N_{failC} = bondline failure, $N = F_{su} \times L$, and

N_{failD} = failure load recorded from experiment.

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lation methods N_{failA} is the anticipated laminate strength based on the Tsai-Wu [12,13] quadratic failure criteria. Both the first ply failure (FPF) and the last ply failure (LPF) strengths are listed. These values are important to compare the joint strength with laminate (adherend) strength. The goal of the bonded joint is for the failure strength of the joint to equal the failure strength of the laminate adherends (after thermal and environmental considerations). If this goal is accomplished, the joint is considered to have an efficiency of 100%. This is also true of a patch repair. The goal in current laminate repair is to restore 100% of the laminate strength. And, in fact, according to Hart-Smith [3], a joint strength of approximately 1.5 times the laminate strength is desirable, incorporating a safety factor to ensure that the relatively unreliable bonded joint retains at least the strength of the more reliable laminate during the life of the structure. This factor incorporates environmental and fatigue considerations into design.

In the second column of Table 2, N_{failB} represents the calculated strength of each specimen, given that failure occurs in a shear mode in the bondline. This strength has been calculated by taking the FE bondline shear stress data for each specimen and extrapolating that data to the load level where the shear stress anywhere along the length of the joint equals F_{su} of the adhesive. This strength calculation would be considered conservative by the current method of calculating joint strengths, because Hart-Smith [3] suggests that a given amount of plastic behavior is desirable and may be unavoidable at high loads.

Looking to the next column in Table 2, N_{failC} is simply $F_{su} \times L_{\text{bondline}}$. This strength is considered nonconservative because it requires that the entire bondline must experience F_{su} before failure occurs. Because N_{failB} is considered conservative and N_{failC} is considered nonconservative, these two values should form an envelope for the tested strength. This proved not to be true, however, because many of the actual failure loads, N_{failD} , are outside of and, in fact, below the envelope values. Only the strengths of Specimens 3 and 4 are within their respective strength envelopes. Specimen 1 is farthest outside the envelope at only 23% of the calculated lower envelope value.

This comparison of calculated and actual strengths of the composite lap joint specimens indicates the inability of the FEM to give adequate information that would lead to the calculation of the strengths of the specimens. While FEM were not able to give information leading to the prediction of failure of the specimens a priori, they were able to indicate the relative strengths of the specimens, and accurately predict which specimens were most and least structurally efficient. This was true except for Specimen 5. The FEM suggested a strength much larger than Specimen 3, but experiment did not bear this out. In general, the model strength predictions were nonconservative, suggesting strength levels that were not attainable during testing.

A significant reason for the inability of the FEM to predict failure lies with the fact that the composite material was not completely modeled. First, each ply property is incorporated into a common laminate. Then, that laminate's properties are put into the FEM to give the adherends their isotropic properties. Failure, however, occurred in all the specimens in a characteristic mixed mode that was not expected and not analyzed for. As was described, a crack that possibly initiates in a 90° ply propagates up through the laminate of the patch adherend and at least contributes to, if not causes, failure. This 90° ply has not been modeled, and so this failure mode is not detectable through the FEM.

The mode of failure analyzed for in this paper is a shear failure mode within the adhesive bondline. This anticipated failure would begin with a crack initiating at the center of the thickness and at the end of the bondline as a result of shear stress rupture and would propagate through the adhesive, leaving the specimen separated into its two constituent adherends. The failure that actually occurs, however, is a crack that may initiate outside of the bondline, possibly in a bondline contacting 90° ply. This crack may also initiate anywhere along the interface of the patch adherend and the film adhesive. Because of all specimens tested, very little adhesive

was found on the patch adherend fracture surface. So, rather than a crack initiating and propagating in the adhesive within the confines of the bonding surfaces of the adherends, the crack, for all specimens tested, propagated in a staircase path, leaving pieces of the patch adherend plies adhered to the parent adherend fracture surface. This failure mode is characteristic of only built-up fiber composite laminate adherends and would not occur in isotropic metallic adherends.

The failure mode of the specimens just described suggests that contributing failure stresses may be different from those of a bonded lap joint in which the adherends are isotropic. Upon inspecting the bondline and the surrounding area of a bonded composite scarf joint, one may note that the bondline interface ply changes with position, x , along the bondline. Because of this change of interfacial ply, the stiffness of the material that interacts with the bondline changes. This, in turn, changes the extent of shear stress transfer into the bondline. An assumption made while performing a stress analysis on an isotropic adherend bonded lap joint is that the adherend material is much stiffer than the film adhesive. Because of this, the adhesive deforms readily, conforming to the extensional characteristics of the adherends. During an isotropic analysis that uses this assumption, the bondline remains a constant thickness, η , ensuring that shear strain

$$\gamma_{\text{bondline}} = \frac{\delta_1 - \delta_2}{\eta}$$

depends on the relative extension of the adherends. In a composite joint, however, the interfacial adherend material cannot be assumed to have a much larger stiffness than the bondline adhesive. An illustration of this fact may be made by observing that the transverse stiffness of a ply of AS4-1634 is $E_y = 8900 \text{ MPa}$ ($1.3 \times 10^6 \text{ psi}$), while the stiffness of Metalbond 329 is $E = 10\,300 \text{ MPa}$ ($1.5 \times 10^6 \text{ psi}$). (These parameters may be viewed as roughly equivalent when they are contrasted to the strain capabilities of each: $\epsilon_y = 0.5\%$ for AS4-1634 versus $\epsilon_{\text{fail}} > 10\%$ for Metalbond 329.) Because of this relative stiffness equivalence, some bondline strain behavior may be conveyed to the compliant 90° plies. The brittle 90° ply then becomes part of the bondline and is forced to adopt the shear strain behavior of the ductile adhesive and undergoes shear distortion of a magnitude comparable to that of the bondline. This forced behavior causes the 90° ply to crack in a shearing mode (Fig. 13). These cracks may be responsible for the tearing of the 90° plies from the patch adherend during the failure process.

Evidence exists from testing the specimens that premature partial failure of the joint occurs, due to strain incompatibility of the bondline contacting interfacial plies. The comparison of the shear strain behaviors of Specimens 3 and 5 at 2.63 KN/cm (1500 lb/in.) (see Fig. 10) suggests that Specimen 5 would have greater strength because at this load the bondline expe-

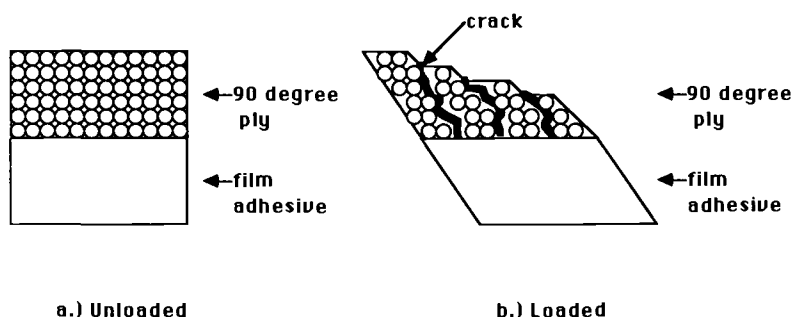


FIG. 13—Cracking in 90° interfacial ply as a result of bondline shear strain.

periences a lower shear strain. In addition, the laminate that comprises the patch adherend for Specimen 5 was approximately twice as strong as the patch adherend for Specimen 3. Despite these factors, Specimen 3 was stronger than 5. Because both adhesive and laminate performance were more favorable for Specimen 5, this suggests that interaction of the adhesive with the laminate caused failure. The stiffer adherend for Specimen 5 was able to conduct load more efficiently into the interior of the joint where the more compliant, and at the same time, more brittle 90° plies made contact with the bondline. While testing this specimen, cracking at approximately one half the failure load was heard, which may be the point at which the 90° plies became disbonded. This is supported by the load-versus-time curve, which showed that load falls off several times during the loading process. In addition to the cracks that may appear in the 90° interfacial plies, cracks that are a result of extensional strain and first ply failure (FPF) of the laminate may appear in the 90° plies at higher loads and may aid in the failure process. The joint failure load for Specimen 5, however, was significantly below the FPF load.

As a result of observing the crack patterns of the failed specimens, questions arise. Why was the patch adherend always swept clean of its adhesive and some of its bondline contacting plies? Why did the crack always initiate and propagate on the patch side of the adhesive? The method of manufacture of the specimens may help to answer these questions. During fabrication of the joint, the cured parent adherend was scarfed, film adhesive was laid, then patch plies were laid one atop another, producing a staircase of plies at the patch adherend-film adhesive interface. This stairstep interface resulted in a variable thickness bondline and produced a patch adherend bonding surface not amenable to bonding. The parent adherend was fully scarfed, on the other hand, producing a bonding surface that was much better than the patch adherend for bonding. When fracture occurred, it was natural for the crack to propagate along the surface that required the least energy for fracture, the surface of the patch adherend that provided a poor bonding surface.

Conclusions and Recommendations

The method used in this paper to increase the efficiency of a composite patch repair has merit and should be pursued. By reorienting plies in the principal stress direction, it is possible to reduce the number of plies and thickness required to form a usable composite repair patch. Because of the additional patch laminate stiffness due to the reoriented plies, joint efficiency is increased. The additional stiffness of the patch reduces the gradient of extensional strain in the patch, resulting in a more uniform shear strain-stress distribution in the bondline when load is applied. Additional work is needed, however, to understand the ply angle and sequence effects on the strength of the joint. Reducing the large range of shear stress along the joint was not adequate to increase the efficiency of the joint described by Specimens 3 and 5.

The method of increasing patch stiffness by reorienting plies should not be used for laminates that experience a stress field where the first and second principal stresses, σ_1 and σ_2 , both have large magnitudes. Because increasing stiffness and strength in the "1" direction causes a decrease of those laminate properties in the "2" direction, premature failure in an unwanted mode may occur if stress in the "2" direction is appreciable.

Strength theories or criteria for isotropic bonded joints are not directly applicable to composite bonded joints and repair patches. This is due to the additional sites available for crack initiation and paths available for crack propagation in the composite material. The sites are predominantly located at the 90° ply-adhesive interface. These sites may also be located at the 45° ply-adhesive interface but are less critical. Effective theories developed for strength predictions must address the dominant failure mode (that is, failure that preys on the weak, brittle 90° ply-adhesive interface). In addition, the strength of the composite repair patch is greatly affected by the bondline contacting "staircase" of plies that results from the layup of a patch

onto the scarfed parent laminate. Each “stairstep” acts as a stress raiser and provides additional locations where failure might initiate. Prescarfing and then application of a precured patch may make the patch bond stronger. However, even with a fully optimized bonded joint, strengths in excess of the last ply failure strength of the laminate may not be attainable.

It is necessary to model the plies of composite laminate in the adherends of bonded composite joint. The bonded joints with composite adherends represented in this paper provided an unacceptable prediction of strength, because the ply weaknesses and the interlaminar interaction of the composite are not modeled. All of the plies of the composite must be modeled and possibly even the interlaminar matrix for the model to predict where fracture is likely to initiate. Work on geometrically and materially nonlinear FEM should be accomplished to try and gain better agreement with tested values, but more importantly, to establish an FE method that might aid in the accuracy of the FEA of the entire aircraft structure that incorporates bonded joints or repairs.

The experimental method used in this paper for testing composite bonded lap joints is adequate for basic experimental stress analysis of bonded composite joints, although testing on composite patched panels should be accomplished to gain a fuller understanding of the interaction between the critical zone where failure is expected to initiate and the rest of the composite repair.

A formula that estimates strength may be generated based on the observed behavior of the tested and analyzed composite bonded joints. For a fully scarfed joint composed of isotropic adherends, the strength may be estimated as

$$N_{iso} = CL^{1/k}$$

where N_{iso} is proportional to the k th root of the overlap length, L . The diminishing increase in strength with increase in L is due, in part, to the fact that as lap length increases, the probability increases that the bondline will have an additional flaw. An example of this is a localized thin bondline that results from excessive processing pressure. This thin bondline will have escalated shearing strains and may provide the location for crack initiation. For a fully scarfed joint free of flaws, $C = F_{su}$, $k = 1$, and

$$N_{iso} = F_{su}L$$

The strength of the joint may be further degraded in the case of a joint composed of composite adherends

$$N_{comp} = N_{iso} \left[1 - A \frac{n_{90}}{n} - B \frac{n_{45}}{n} - C \frac{n_0}{n} - D \frac{t_{step}}{t_{lam}} \right]$$

where the ratios n_{90}/n , n_{45}/n , n_0/n represent the fraction of the bondline affected by the strength loss. The constants A , B , and C range from 0 to 1.0 and represent the magnitude of the strength loss. For example, $n_{90}/n = 0.25$ for $\pi/4$ quasi-isotropic laminates, and $A = 1.0$ because 100% of the lap length in the adherend interface, which are 90° plies, lends no strength to the composite joint. t_{step}/t_{lam} is a ratio of the step thickness divided by the thickness of the laminate and degenerates the calculated joint strength based on the severity of the step stress raiser.

Additional work is necessary to develop FEM that can predict joint failure reliability. Also, joints of higher efficiency than already exist must be developed so that repairs of composite structure can restore all of the strength lost from damage. New joint designs may be used that lower the high bondline shearing strains at the end of the bondline and in the vicinity of the

line at the critical locations may reduce bondline shear strains enough to inhibit the shear cracking of the bondline contacting 90° plies. In addition, entire patched composite panels must be tested to validate the efficiency of the repair and the reliability of the strength analysis.

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ISBN 0-8031-1426-5