The Effect of Interlayers on the Mechanical Response of Composite Laminates Subjected to In-Plane Loading Conditions

by

Robert Edward Swain III

Thesis submitted to the Faculty of the Virginia Polytechnic Institute and State University in partial fulfillment of the requirements for the degree of Master of Science in Engineering Mechanics

APPROVED:

__________________________
Kenneth L. Reifsnider, Chairman

__________________________
Wayne W. Stinchcomb

__________________________
John C. Duke, Jr.

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(ABSTRACT)

"Interlayering" — the incorporation of low-modulus film adhesive between the plies of composite laminates — has proved to be a successful technique for reducing debilitating out-of-plane stresses. This work seeks to determine the effect interlayering has on a composite laminate's in-plane performance.

Two laminate systems, an unnotched, 16-ply, quasi-isotropic, AS4/C985 and a center-notched, 32-ply, quasi-isotropic, AS4/C1808, were furnished in an interlayered and baseline (non-interlayered) configuration. The interlayers, 0.0005 in. each in thickness, appeared between each ply in every laminate tested. Both configurations of these two material systems were subjected to a regimen of in-plane loading tests. These tests included monotonic tension and compression, fully-reversed (R = -1), tension-compression fatigue cycling, and long-term tensile loading. A new test method, called the Incremental Strain Test (IST), was developed in an attempt to isolate and distinguish the long-term, tensile response of the interlayered and baseline laminates. This technique and its utility are described herein.

The interlayered laminates exhibited superior performance during monotonic and IST loading. Distinctly higher ultimate loads and strains were achieved by the interlayered laminates. The notched fatigue performance of the interlayered laminates was sub-standard in comparison to the baseline results at the load level tested. The residual tensile strength of the fatigued interlayered laminates fell sharply at an early fraction of the laminates' total life. The presence of the interlayers did not degrade the laminates' IST performance.
Several non-destructive techniques were used to monitor the damage mechanisms. These results, when combined with the experimental findings, helped explicate the disparity found between the interlayered and baseline laminate response.
Dedication

This work is dedicated to my wife,
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Introduction and Literature Review

The merit and utility of composite materials as structural components in existing and future designs is unquestionable. Products are introduced daily into the market with composites replacing conventional engineering materials. Composite materials free the designer from material constraints allowing him, instead, to "tailor" the material to the design constraints.

Ideal composite laminate response is modeled using classical lamination theory. Any deviation from this ideal, i.e., the introduction of damage into the laminate, significantly alters the predicted response. Over the last decade an intensive effort has focused on modeling damage and damage growth in laminates in order to realistically predict their mechanical behavior. Some success has been achieved in identifying specific damage modes, e.g., delamination [1] and matrix cracking [2], and their effect on stiffness degradation. An elusive, yet, fundamental goal, however, is the understanding of damage/strength relationships. Today, it is still unclear how damage affects laminate strength. A successful methodology, however, may be replicated. By isolating specific damage modes in degree and extent, and then performing residual strength testing, valuable information may be obtained. Advanced designs cannot depend on our models of ideal behavior; we must model the real, physical situation for prediction and knowledge sake.
Interlayered Material

To compensate for our incomplete understanding of damaged response, material engineers have espoused the age-old principle, "beef it up where it breaks." Where damage modes have been identified, engineers have taken steps to prevent or avert such damage. Epoxy systems have been created with the ability to "heal" after cracking, given the right conditions [3]. Fiber/matrix interfaces have been treated and/or coated with external agents to impede interfacial failure [4-6]. It is the prevention of delamination — termed, "the most prevalent life-limiting mode in advanced composite structures," [7] — that has received the most attention.

Early studies of delamination indicated the dependency of interlaminar stresses on stacking sequence [8]. Careful re-arrangement of the stacking sequence could sufficiently reduce or reverse the magnitude of these out-of-plane stresses [9], yet, would prove futile if the load excursion was reversed. Attempts have been made to reinforce laminates in the out-of-plane direction at sites where delamination is found to initiate. Huang, et al. [10], employed steel wire as an edge reinforcement in the thickness direction. Embedding the wires by hand proved impractical despite the resulting reduction in delamination. Mignery, et al. [11], sought to improve upon this idea by using an industrial sewing machine to stitch Kevlar fibers into the thickness. The stitching successfully contained delamination, however, it proved useless in preventing its initiation. The effect delamination suppression had on the ultimate stress was inconclusive. Edge stitching produced a 25% increase in strength for a "matrix-dominated" laminate, however, a decrease in strength was reported for quasi-isotropic and "fiber-dominated" laminates. Kim [12] reinforced the free-edge with fiberglass woven cloth using a structural adhesive. The delamination threshold was reported to increase 100% and more for laminates susceptible to edge delamination. He noted, however, that "the reinforcing process, including selection of reinforcing material and adhesive, must be further improved to achieve a more reliable reinforcement."
Within the last ten years a new approach to the delamination problem has surfaced. Principles of linear elastic fracture mechanics have been applied to the initiation and propagation of delamination. Work by Rybicki, et al. [13], Wilkins, et al. [14], and O'Brien [1] have demonstrated the use of the strain energy release rate, \(G\), as a critical parameter in the initiation and propagation of delamination. An excellent review of their work may be found in [5]. By increasing the material's critical value of \(G\) — known as the critical strain energy release rate, \(G_c\) (or, proportionally, the fracture toughness) — along the crack path, much more energy is required to propagate a delamination, and therefore delamination damage may be retarded. Devitt, et al. [15] and Whitney, et al. [16] have reported utilizing the double cantilever beam (DCB) test as a method of measuring the interlaminar \(G_c\) due to mode I cracking. Materials tested as having a high \(G_c\) would seem to be excellent candidates for "buffer strips", i.e., material placed at delamination-critical sites in order to control delamination growth. This reasoning is the motivation behind the evolution of interlayered composite materials.

Lackman and Pagano [17] first introduced the notion of "softening" the free-edge with their "mechanical serration" scheme. The free-edges of the 90° plies were removed through serration in a boron/epoxy laminate with a \([\{\pm 30\}_2/90_2/(\mp 30)_2]_T\) lay-up, and replaced with epoxy-resin in order to reduce the interlaminar normal stresses. While a 30% increase in static ultimate strength over the baseline material was found, the fatigue response was severely sub-standard. Notions of reducing the 90° width by the boundary layer distance and filling this void with resin were not tested experimentally.

In April 1984, three independent research teams presented work performed on laminates containing discrete "inner layers" of toughened material seeking to improve delamination resistance. Browning and Schwartz [7] employed the DCB test to measure the \(G_c\) of various high-fracture toughness materials co-cured in the midplane of a unidirectional AS1/3502 laminate. A combination of adhesive and Kevlar matting proved the most fracture resistant. The \(G_c\) of this laminate was 12.5 times higher than the baseline (non-interlayered) laminate.

Krieger [18] outlined the developmental motivation behind interlayered composites laminates. In an effort to increase laminate toughness, he identified the epoxy resin's ultimate
shear strain as a key parameter. Experience related, however, that resin with good toughness usually exhibits poor hot/wet compressive strength. Development of a new epoxy resin yielded both adequate toughness and compressive strength, yet, the compression strength after impact (CSAI) was found to be disappointing. From this, came the idea of combining the resin roles synergistically. Tough resin with good compressive strength could serve as the fiber binder while a thin, discrete layer of ultra-tough resin exhibiting superior ultimate shear strain could absorb and dissipate impact energy. Experimental tests on Cycom HST-7, a "duplex matrix" system (see Figure 1 — an analogous material system), confirmed these postulates. Krieger pointed out two major advantages of the interlayer concept. Cracks intersecting an interlayer will have a reduced shear stress concentration when compared to a mono-matrix system. Similarly, the interlayer, with its large strain to failure, will not easily crack. Therefore, high strain in critical elements may be attained without local, regular, strain (and stress) concentrations due to adjacent off-axis cracking.

Chan, et al. [19], incorporated a thin (0.005 in.) layer of tough adhesive (American Cyanamid FM 1000) into the critical interfaces of an AS4/3501 laminate with a [(±30)_2/90_2/(±30)_2]_T lay-up — the lay-up used in the edge delamination test [1]. These laminates were subjected to a regimen of static tests. They concluded from this study, that:

1. Edge delamination in these AS4/3501/FM 1000 laminates was suppressed until final failure.
2. This suppression was attributed to the reduction of the percentage of $G_i$ in the total strain energy release rate due to the addition of adhesive layers.
3. Interlaminar normal and shear stresses were reduced at these critical interfaces.

Subsequent research on composite interlayer systems has focused mainly on their ability to resist delamination in laminates and loading conditions where delamination is a prevalent damage mode. Tests on interlayered composite laminates subjected to static loading [20,21],
Figure 1. Edge photomicrograph of an interlayered material system with damage.
fatigue loading [22], multi-span shear loading [23], and impact [5,24-29], have shown that the interlayer has performed its task remarkably well. Improving the interlayer material in order to optimize the host laminates' Compression Strength After Impact (CSAI) is the goal of the work by Evans and Masters [24], Masters [25], and Hirshbuehler [26]. Recently, an analytical model for an intersecting and interfacial crack in a laminated structure containing interlayers has been proposed by Kaw and Goree [30] as a modification of the work of Gecit and Erdogan [31]. Each analysis examines the stress concentrations produced from the cracked plies as a function of interlayer thickness.

A review of the literature reveals that the term "interleafing" has become the conventional description of the new material lay-up technique introduced above. The term "interlayering" and its respective conjugations are chosen over the former term in this work, not out of disrespect, but for these two reasons:

1. "Interlayer" embodies the term "layer", which accurately describes the fact that discrete "layers" are interspersed between individual lamina. "Interleaf" — while an appropriate term — is historically used in the context of books and pages.

2. Experience has shown that the term "interlayer", perhaps due to the reason stated above, allows the layman to readily conceptualize the physical situation.

**Problem Statement**

The literature is replete with data espousing the merit of interlayered composite design as a means to significantly increase delamination and impact resistance [5,6,18-29]. What is yet to be substantially explored is the effect interlayering has on a composite laminate's in-service mechanical response. Newaz, in [24], poses the following question:
In the case of long-term loading of the laminate in actual application, I would be very concerned about the "interleaf" material from the standpoint of creep. Although one may improve delamination toughness of the composite, can you convince me why one should consider such laminate design for overall long-term performance?

One should not only be suspicious of long-term response, but, with the profound effect the interlayer has on the local, three-dimensional stress state, one should wonder if the interlayer markedly alters the monotonic and fatigue response. It is the goal of this investigation to present relevant data and submit a cogent discussion of the significance of the results pertaining to the mechanical response of interlayered composite laminates subjected to the load conditions mentioned above.

**Interlayer Response to Mechanical Loading**

Having established with convincing proof that interlayers do suppress delamination initiation and growth, investigators have set out to assess the effect this configuration has on the laminate response under various loading conditions. Their motivation to perform such work was spelled out earlier in the introduction. If one can isolate a damage mode, i.e., free-edge delamination, 90° matrix cracking, etc., stress analysis and strength testing may be used to determine the effect this damage has on the stress distribution and, subsequently, on laminate strength. Such data is essential in order to completely understand damage/strength interaction.

**Monotonic Response**

Chan, et al. [19], tested non-interlayered and interlayered laminates to static failure. They reported that the interlayers had suppressed delamination until final failure. With this data,
one might extrapolate the role delamination plays on strength reduction. Whole adhesive layers and strips of adhesive placed on the edge were co-cured into the critical interfaces of two laminate systems; a "matrix-dominated" laminate ([\(\pm 30\)\_\(\L_90\)/\(\L_{\pm 30}\)_\(\L_90\)]) and a "fiber-dominated" laminate ([\(\pm 35/0/90\)_\(\L_90\)]. Ultimate strength of the "whole" interlayered "matrix-dominated" laminates rose 24% from their baseline value. An increase in ultimate strength for the interlayered "fiber-dominated" laminates was reported for each configuration. Strength rose 12% for laminates with "whole" adhesive layers while in an increase of 16% was reported for strips placed along the free edge. They concluded that "an adhesive layer placed at critical locations in laminates has successfully been demonstrated to improve the laminate strength." They also noted that the introduction of an interlayer "adjacent to a 90° ply provides a larger shear lag zone and consequently reduces the 90° ply crack density."

In a later work, Chan [22] experimented solely with strips of an adhesive layer placed on or interior to the free edge. Strips placed interior to the free edge had no positive effect on ultimate strength for their "matrix-dominated" laminates (a 4% reduction in strength was indicated). Yet, a stunning increase (32%) in ultimate strength was reported for this configuration in the "fiber-dominated" laminates.

Soni and Kim [20] interlaid thin layers of Cycom HST-7 into T300/1034C laminates. They investigated a [\(0/\pm 45/90\)_\(\L_90\)] lay-up and a [(\(\pm 30\)\_\(\L_90\)] lay-up with and without an interlayer at the mid-plane, with the additional case of an interlayer between the 90° ply and the respective off-axis ply. This latter case proved the strongest. An increase of 15% in ultimate strength over the baseline [\(0/\pm 45/90\)_\(\L_90\)] laminate occurred, while a 29% increase was found over the baseline [(\(\pm 30\)\_\(\L_90\)] laminates.

In each of the studies above, interlayers were selectively incorporated into laminates, especially at interfaces prone to free-edge delamination due to high interlaminar normal stresses. In all cases, delamination at these interfaces was suppressed and, for the most part, causatively or not, ultimate strength was increased. While certain empirical studies [32] have attempted to assess the effect free-edge delamination has on static ultimate strength, others have approached the problem analytically [32,33].
In contrast to the above work, Sun and Jen [34] have sought to isolate the effect of matrix cracks — not free-edge delamination — on laminate strength. To this end, 5 mil thick layers of American Cyanamid FM 1000 adhesive film were co-cured into the laminates, isolating the 0° plies in a [0/90°/0] and a [90°/0°/90°] lay-up. The following observations were reported:

1. The adhesive layers seem to diffuse the stress concentrations in the 0° plies caused by the adjacent 90° ply cracking.

2. By reducing the 90° crack density, fewer stress concentrations of less magnitude affect the 0° plies.

3. Matrix crack-induced interior delamination in these cross-ply laminates was suppressed.

4. In conclusion, all of the above factors contributed to the 30% increase in laminate ultimate strength over a non-interlayered system.

The work cited above provides valuable data to those analysts, who, for several years now, have been studying the effect matrix cracking has on the laminate stress state. Talug [35] has shown using the finite-difference method that local stress concentrations arise in plies adjacent to off-axis matrix cracks. Jamison [36] has presented convincing proof showing the effect these concentrations have on adjacent ply fiber failure. Electron microscopy has revealed fiber failures in statically loaded [0/90°]s graphite-epoxy laminates to be concentrated in narrow bands that correspond to the adjacent 90° matrix cracks. Work by Reifsnider and Talug [37], Crossman, et al. [38], and Strauss [39] have concluded that matrix cracks intersecting laminate interfaces create interlaminar shear and normal stresses sufficient, in many cases, to initiate interior local delamination. Highsmith and Reifsnider [40,41] investigated this notion further, noting that these matrix crack-induced delaminations cause further stress (and strain) redistribution and may, themselves, promote fiber failure and subsequent laminate failure. In [42], O'Brien concludes:

Local delaminations growing from matrix cracks, however, create local strain concentrations that may lead to nominal laminate failure strains below the in situ failure strain of the primary load-bearing plies.

Introduction and Literature Review

9
Fatigue Response

Unnotched

In a previously cited work, Chan [22] employed adhesive strips along and interior to the free-edge at the critical interfaces of a "matrix-dominated" and "fiber-dominated" laminate to investigate delamination arresting. Unnotched baseline and interlayered laminates were subjected to tension-tension (R=0.1) fatigue loading with the maximum amplitude equal to 90% of the static delamination threshold strength. Delamination was successfully arrested by the interior adhesive strips and completely suppressed by the edge strips. Residual tensile strength tests were performed for each laminate and each configuration upon reaching 4 million cycles. Some of the more interesting findings from this study are summarized below:

1. The baseline "matrix-dominated" laminate could not survive the load amplitude past 2.6 million cycles.
2. With interior strips interlaid into this laminate, the residual strength at 4 million cycles decreased 15% below its static laminate strength. Upon lowering the load amplitude 20%, the residual strength at 4 million cycles rose 13% over its static laminate strength.
3. In the baseline "fiber-dominated" laminate, the residual strength at 4 million cycles rose 11% over its static strength.
4. At 4 million cycles, the interior strip configuration for this laminate revealed only a 7% increase in residual strength over the baseline residual strength while posting an 11% decrease versus its own static strength.
5. The edge strip configuration had a 7% increase in residual strength over the baseline figure and no significant change compared to its own static strength.

**Notched**

To the author's knowledge, no study to date has been published documenting the effect of interlayering on notched fatigue response. Yet, one might extrapolate from the results of a recent work. Simonds, et al. [43], investigated the notched fatigue response of quasi-isotropic AS4/poly(etheretherketone) (PEEK) laminates subjected to tension-compression (R = -1) fatigue. PEEK is considered a "tough" thermoplastic matrix system since its critical strain energy release rate, $G_c$, in both mode I and mode II cracking, is at least twice as large as conventional epoxy systems [44]. On this basis alone — several studies would indicate — delamination resistance is greatly heightened. In a thermoplastic fracture model proposed by Su [45], the extensive deformation occurring in the crack-tip process zone accounts for the large amount of energy dissipated in crack propagation. Thus, it is the qualities of delamination resistance and highly plastic (matrix) crack-tip zones in PEEK that make it comparable to an interlayered laminate. Summarizing the notable results from [43]:

1. Fundamental to the complex fatigue response of center-hole notched laminates is the on-going competition between interacting damage modes. Matrix cracking, delamination, and fiber failure near the hole tend to reduce global geometric stress concentrations while simultaneously inducing local stress concentrations [46].

2. At low cyclic stress amplitudes, delamination growth predominates throughout the life of the AS4/PEEK laminates; a life which exceeds that of T300/5208 cycled under like conditions. Compression/instability failures are the rule for each AS4/PEEK laminate at this load level.
3. At high cyclic stress amplitudes, damage confines itself to a narrow band emanating from the hole across the width of the specimen. Fiber fracture is prevalent across this damage band resulting in tensile failures premature in life to its T300/5208 counterpart. Unlike the AS4/PEEK, the T300/5208 laminates do not reveal a transition in failure mode due to the higher load level.

4. Under cyclic loading conditions, improved matrix toughness does not necessarily translate into improved mechanical performance.

The tendencies observed above will serve as a qualitative and comparative measure of the notched fatigue response of the interlayered laminates in this study.

**Damage Tolerance and Durability Concepts**

The interlayering concept was conceived and developed by many researchers in order to improve the damage tolerance of composite materials. The term "damage tolerance" is quite nebulous. Since interlayering is predicated on the concept of "damage tolerance", a brief discussion of its definition and measurement techniques is included in this review.

**Definition**

In [47] the following definition for "damage tolerance" appears:

*Damage Tolerance — The ability of a structure to resist failure due to the presence of flaws, cracks or damage for a specified period of time.*
Griffin [48] echoes the gist of this definition, qualifying that “damage tolerance” — for the case of nonvisible impact — means that “structures must be able to carry design ultimate load without failure for the service life of the structure.” Often, “damage tolerance” is confused with the oft-used term “durability”. Compare the definition of “durability” in [47]:

Durability — The ability of the structure to resist structural degradation due to moisture, thermal effects and normal usage.

“Structural degradation” should be viewed as a structure’s loss of stiffness, strength, and life. “Damage” inception and propagation is the process which directly dictates the level of structural degradation. Damage growth (where “growth” encompasses both inception and propagation) is monitored in composite laminates through stiffness measurements. A loss in laminate stiffness indicates the presence of damage growth. Unfortunately, no simple qualitative relationship exists between damage growth (stiffness loss) and residual strength. Therefore, monitoring stiffness loss is the most straightforward and simplistic approach toward quantifying “structural degradation”. A crucial difference in the two definitions above may now be highlighted. “Damage tolerance” is a measure of resistance to failure given that damage is present. “Durability” is a measure of resistance to stiffness loss. With this distinction in mind, a review of damage tolerance tests is made.

Test Techniques

The measurement of resistance to failure is universally deemed “strength”. Therefore, a damage tolerance test is often simply a strength test after damage is introduced. Deciding what damage should be considered and to what extent this damage is introduced is a crucial problem, however. In composites many damage modes exist, and rarely do they exist alone. To circumvent this formidable problem, some researchers have viewed “damage tolerance” as being synonymous with “damage containment” [18]. Given that damage exists only in
subcritical elements [49], by effectively "containing" the damage to these elements, no appreciable loss of strength or life should occur. With this rationale, material tests are sought to quantify this ability to contain damage.

Motivated by the work of Byers [50], many researchers [5,18,24-26,28,29,48,51] employ the Compression Strength After Impact (CSAI) test to measure comparative damage tolerance. Conventionally, a drop weight is used to impart a 1500 in-lb/in impact to a fixed panel. The as-damaged panel is then loaded in compression until failure occurs. This test measures a material's durability first, then the damage tolerance as a function of its inherent durability. The test is used as a comparative, not absolute, measure.

A couple of shortcomings in this test method come quickly to mind. First, low-velocity impact is a thoroughly severe damage initiator [52] causing widespread delamination that greatly reduces compressive strength. Perhaps the best damage tolerance test would be one in which the material was exposed to probable in-service damage and then subjected to typical load conditions (i.e., tensile, compressive, fatigue loading, etc.). Impact loading — a worst case scenario — may not represent the physical situation very well. Secondly, if one wishes to strictly adhere to the definition of "damage tolerance" given, time-at-load is not accounted for in this test method. Not only must the material be able to support the ultimate design load, it must support it over time under service conditions.

Any reliable test method that duly incorporates the effect of time is, by nature, long-term. Usually, the longer the test, the more credulous the results, since they best typify the actual long-term condition. Yet, such long-term testing is too costly, both temporally and financially. For many years, a huge thrust toward developing "accelerated" testing has been under way. Such testing attempts to represent the actual long-term behavior precisely, yet, accomplish this feat in a fraction of the time.

Epoxy matrix systems are viscoelastic materials, i.e., their response is highly time-dependent. Any laminate loaded in a "matrix-dominated" direction will reveal its viscoelastic nature. For this very reason, creep testing — perhaps the most common measure of durability — is quite an intensive area of research in composites. Moreover, "accelerated" creep testing...
is receiving a considerable amount of effort and attention [53-55]. The focus of this research, so far, has been on unidirectional "matrix-dominated" laminates. Multi-directional laminates, especially those with fibers in the load direction, are rarely tested since most fibers are highly resistant to large creep deformation. Yet, it is well known that off-axis matrix damage does influence load-carrying plies through local stress (and strain) concentrations and global load transfer [34, 42, 56, 57]. Therefore, a viable, "accelerated", long-term durability test method is needed that might highlight the effect of any incipient off-axis damage on the long-term response.

Such was the motivation behind the test method introduced by Vittoser and Reifsnider in [58]. Called the Incremental Loading Test (ILT), the method involves loading to a constant strain which is held for a fixed period of time, allowing stress relaxation to occur. When the time period has elapsed, an increment of strain is added and the process is repeated (see Figure 2). This simple loading scheme is not new. Researchers of viscoelastic materials know it as "interrupted-ramp strain input," a common test for homogeneity — the necessary condition of material linearity [59, 60]. Mecklenburg and Evans [61] have employed this loading scheme in order to obtain an "equilibrium" stress-strain diagram for various adhesives. What is considered "new" about this test method is its application to composite laminates and the valuable information obtained from the response.

The ILT will be presented (with slight modifications) in this study as an effective alternative to the CSAI. Like the CSAI, the ILT measures a laminate's damage tolerance as a function of its durability. Contrary to the CSAI, however, it allows one to systematically control the level of damage input. The test also measures laminate response as a function of time; conforming to the given definition of "damage tolerance". The ILT serves as a valuable, "accelerated", durability and damage tolerance test.
Figure 2. Schematic of IST loading — Strain vs. Time.
Objectives

The primary objectives of this present study are summarized below:

1. To determine the effect interlayering has on the damage mechanisms within a composite laminate. This is accomplished by comparatively subjecting interlayered and non-interlayered laminates to different in-plane loading conditions while assessing the damage development in each respective laminate.

2. To understand how the damage mechanisms in interlayered composites affect the mechanical behavior of the laminate. Put in simpler terms, could the presence of the interlayer enhance (or degrade) the monotonic response? the fatigue response? the long-term response?

3. To introduce the Incremental Strain Test (IST) — a modification of the ILT — by describing the new test technique and presenting its results. It is the intention of the author to provide evidence that the IST is a valuable test in the investigation of damage development in composite laminates.
Experimental Investigation

Materials and Test Specimens

The investigation described herein examines three different laminated pre-preg systems. These systems and their laminates are described below.

AS4/Cycom 985

Pre-preg tape comprised of Hercules' AS4 fiber bound in American Cyanamid's Cycom 985 matrix was laid-up into several 16-ply, quasi-isotropic panels by American Cyanamid. AS4 fiber — a long-time industry standard — possesses a 1.5% strain to failure [5], yet, often finds its potential under-utilized by brittle matrix systems. Cycom 985, however, presents a necessary improvement. While possessing a comparable tensile modulus to brittle, first-generation systems, increased toughness has allowed for greater durability and damage tolerance. This delicate combination produces a desirable structural epoxy. In [24, 26, 51], the resin and composite properties of C985 are compared to other popular and advanced epoxy systems.
The panels provided were laid-up in two distinct configurations. The baseline panel had a quasi-isotropic, \([0/45/90/-45]\) lay-up. The other panel possessed the identical orientation, however, thin adhesive interlayers, measuring 0.5 mil each in thickness, were co-cured into the panel between each interface, creating a \([0/1/45/1/90/1/-45/1]\) lay-up (see Figure 1). It should be noted that an interlayer did appear between the symmetric 0° plies and adjacent to one of the outer plies. The interlayer is a thermoplastic material designated, for proprietary reasons, as Film C. The integrity of each as-received panel (all panels described herein) was verified using ultrasonic C-scan. Throughout this study, baseline (or, non-interlayered) laminates will be referred to as “B” laminates while interlayered laminates will be designated with an “I”.

Unnotched specimens measuring a nominal 1 in. wide and 5.5 in. long were cut using a water-cooled diamond wheel from the B and I panels. On the average, the thickness of the B material was 0.0915 in. while the addition of the interlayer increased the thickness of the I material to 0.0973 in. The thickness and width of every coupon was measured using a dial caliper sensitive to 0.0001 in. Three specimens of each laminate type were weighed with a scale sensitive to 0.0001 grams. On average, the mass of the I laminates was 6% greater than the B laminates.

Aluminum V-notched tabs were adhered to each specimen with silicone rubber enabling the knife-edges of a strain extensometer to rest within the grooves (see Figure 3). A gage length of 1 in. was consistently used for these specimens. It should be mentioned that all specimens presented herein were stored in ambient laboratory temperature and humidity.

**AS4/Cycom 1808**

American Cyanamid’s Cycom 1808 represents an improvement in resin toughness over their C985. Good comparative data is found for these two resins for neat and composite performance in [24,26,51]. C1808 is shown to possess a larger neat resin flexure failure strain...
Figure 3. An extensometer supported by rubber bands resting within the tabs.
than C985, translating into a noticeable increase in Composite Strength After Impact (CSAI)—earlier mentioned as a standard test of damage tolerance. Panels of 32-ply, quasi-isotropic, AS4/C1808 were fabricated and provided by American Cyanamid. As was the case for the AS4/C985, this material arrived in a baseline and interlayered laminate. The baseline lay-up was \([0/45/90/-45]\) with the interlayered configuration differing only in the addition of the 0.5 mil interlayer at each interface. The interlayered material was, again, a thermoplastic, deemed Film C. Results from toughness tests performed on unidirectional AS4/C1808/Film C laminates are found in [25]. Other AS4/C1808 toughness data are located in [24,26,28,29,51].

Specimens from the I laminate were cut into coupons measuring 1.5 in. wide and 6.5 in. long. A 0.375 in. center-hole notch was drilled in each specimen using a diamond-tipped core drill. The average thickness was 0.1979 in. Aluminum V-notch tabs were affixed across the hole with a gage length of 1 in. (see Figure 4).

Specimens from the B laminates were used in a wholly different study by another investigator (C. E. Bakis). The geometry of the I specimens were purposely matched to the geometry of the B specimens with two exceptions. The length of the B coupons was nominally 6 in. while the average thickness was 0.1810 in. The width of the aluminum tabs differed in the B coupons in an attempt to reduce the local constraint produced by the tabs.

**AS4/3502**

Hercules’ 3502 epoxy resin is a slight modification of its 3501 and 3501-6 resins. AS4/3502 is considered a first-generation pre-preg system having thousands of hours in-service performance. It is routinely used as a baseline comparator with the hopes that the fiber/matrix systems can maintain its strength yet improve upon its poor durability and damage tolerance. The AS4/3502 material in this study is cast in its familiar role as a baseline performer. Comparative composite toughness data between AS4/3502 and AS4/C1808 are located in [25,28,29].
Figure 4. Center-notched AS4/C1808 specimen (with damage).
Pre-preg tape was hand cut, layed-up, and cured on-site at Virginia Tech. The 8-ply, quasi-isotropic panel had \([0/45/90/-45]_8\) lay-up. Coupons were cut from the panel measuring 1 in. wide by 5.75 in. long. The nominal thickness of the specimens was 0.0390 in. No inter-layering was attempted with this material. Again, aluminum V-notch tabs were adhered to the specimens with silicone rubber, each with a 1 in. gage length.

**Mechanical Testing**

All of the mechanical tests described below were performed on a 20 kip (89 kN) electro-hydraulic servo-controlled load frame with hydraulic wedge grips. Common to each test is the gripping procedure. One layer of 180 grit sandpaper (grit side facing out) is affixed with masking tape to the laminate to protect the gripping area from damage caused by the serrated wedge grips. A grip pressure is chosen to ensure no slipping during the complete load excursion while attempting to prevent crushing of the grip area. An ideal grip pressure would cause slipping just above the failure load. Care is taken to ensure that the specimen, as well as the grips, are aligned so that no shear loading occurs.

**Monotonic Tension**

A measure of a laminate's virgin stiffness and strength is essential to the characterization of the material. Such tests were performed on at least two specimens of each laminate type described above. Extensometers were used in all tests to measure strain. The load applied was measured through the testing machine’s load cell. Both load and strain were recorded on an \(X - Y\) plotter.
The tension tests were performed in load control. The control voltage, produced from a function generator, remained linear until failure occurred. A ramping rate of 10 kip/min was chosen for all monotonic tension testing. A peak reader was used to capture the maximum load applied. Stiffness measurements were obtained from the initial linear portion of the load-strain plots knowing the specimen's cross-sectional area.

Monotonic compression tests were performed only on the 32-ply, center-notched, AS4/1808, interlayered laminates (to compare to the baseline results). An unsupported length of 2.5 in. was chosen to prevent buckling failure while maintaining a unconstrained, natural, compressive failure.

**Fatigue Loading**

Fully-reversed (R=-1), tension-compression fatigue was performed on the AS4/C1808 baseline material by C.E. Bakis. The focus of his investigation is a high-load, low-load comparison. The high-load level (a level which induces failure at approximately 10,000 cycles) was considered to be 65% (8000 lb.) of the B material's ultimate compressive strength. The low-load level (the lowest level inducing failure before run-out occurs) was considered to be 50% (6200 lb.) of the B material's ultimate compressive strength. All tests were run in load control with a 10 Hz sinusoidal load applied to the specimen. Strain was measured across the hole with an extensometer. Tests were interrupted periodically to perform non-destructive testing.

The testing procedure for the I material was purposely modeled after the above procedure. A load level of 60% (7300 lb.) of the I material's ultimate compressive strength was chosen, however, to nestle in between the high and low load levels chosen for the B material. Knowing from the Bakis study that the two load levels produce vastly different responses in the B material, testing a modified material at a "middle" load level would allow one to deter-
mine the effect interlayering has on the response by comparing it to the two extremes. Except for the difference in load level, all other test parameters for the two materials were the same.

**Incremental Strain**

Two test procedures were presented in [58], one being a constant load-over-time test, the other a constant strain-over-time test. This study focuses only on the latter test and attempts to differentiate it from the former in both concept and name.

Stressing a material with cracks perpendicular to the applied constant load causes the damage to self-perpetuate. As cracking occurs, axial compliance increases, a reduction in the net cross-section (the area that is carrying load) results, global stress (and strain) increases, more damage occurs, etc., cascading the whole process to failure. At constant strain, however, when damage occurs, the stiffness decreases and the load decreases and will continue to do so until the material can maintain it. The tendency is for the applied stress to decrease at constant strain until equilibrium is met. In the former instance, increased stress results in increased strain, creating an unstable condition. Clearly, loading to constant strain would be very beneficial in trying to "capture" subcritical damage states close to critical (global) failure. Therefore, only "interrupted-ramp strain input" will be used.

Unique to the test method presented in [58,62] is the manner in which the test variables are displayed. In [58,62] the data obtained is plotted as the magnitude of load (stress) relaxation versus constant applied strain (see Figure 5). As the level of constant strain increases, the subsequent stress relaxation response is very non-monotonic. In a multi-directional laminate, "peaks" in this response correspond to off-axis ply damage. Ply damage results in reduced laminate stiffness. If a constant strain is applied while the stiffness is dropping, the stress (load) will relax. The relative magnitude of the relaxation due to this damage is proportional to the reduction in laminate stiffness due to this same damage. Knowing a priori the lay-up of a laminate, one can correlate the successive failure of low-strength, off-axis lamina...
to the sequential occurrence of these "peaks". In the 8-ply, quasi-isotropic, [0/45/90/-45],
AS4/3502 laminate, one would expect that as the applied strain is increased, first-ply failure
(FPF) would occur in the 90° plies. As the strain increases beyond this point, one would expect
the -45° plies to fail before the +45° plies due to the reduced constraint of the failed sur-
rounding 90° plies. With strain still increasing, one would expect the +45° plies to fail next,
since their failure strain in the load-axis direction is less than the failure strain of the 0° plies.
By examining Figure 5, one can see three distinct "peaks" corresponding to the failure of
these off-axis plies. X-ray radiographs taken after each peak confirm the presence of these
ply failures [58].

The graphical representation of the data as displayed in [58] (Figure 5) and [62] has two
shortcomings. The ordinate of the graph misrepresents the significance of the stress relaxa-
tion occurring at the high levels of strain. The ordinate, \( \frac{\Delta \sigma}{\Delta \varepsilon} \), is simply the magnitude of load
drop divided by the original cross-sectional area, normalized by a constant \( \Delta \varepsilon = 500 \ \mu \varepsilon \).
Therefore, the ordinate is directly proportional to the load drop occurring at each constant-
held strain. With the strain held constant at 1000 \( \mu \varepsilon \), a load drop of 50 lb. should be much more
significant than a drop of 50 lb. at 10000 \( \mu \varepsilon \). The applied strain during the IST is expressed,
in mathematical terms, as:

\[
\varepsilon' = (\Delta \varepsilon)(i) + \varepsilon_0, \quad (2.1)
\]

where,

\( \Delta \varepsilon \) is the magnitude of the strain increment,

\( i \) is the number of the increment, and,

\( \varepsilon_0 \) is the strain applied at time, \( t = 0 \).

Therefore, the applied laminate stress is:

\[
\Delta \sigma' = (\Delta \varepsilon')(\varepsilon'), \quad (2.2)
\]

where,

\( \Delta \varepsilon' \) is the total change in the secant modulus during each in-
crement \( i \), and.
Figure 5. Graphical output from ILT (from [58]).
\( \Delta \sigma^i \) is the change in applied laminate stress at each increment \( i \).

Plugging (2.1) into (2.2) and dividing through by \( \Delta \varepsilon \) produces the ordinate:

\[
\frac{\Delta \sigma^i}{\Delta \varepsilon} = \Delta \overline{\varepsilon}^i (i + C),
\]

where,

\( C \) is the constant \( \frac{\varepsilon^o}{\Delta \varepsilon} \).

From (2.3), one realizes that \( \frac{\Delta \sigma^i}{\Delta \varepsilon} \) measures the product of \( \Delta \overline{\varepsilon}^i \) and the current increment number. When \( i = 1 \) (and \( C = 0 \)),

\[
\frac{\Delta \sigma^1}{\Delta \varepsilon} = \Delta \overline{\varepsilon}^1,
\]

yet, when \( i = 10 \),

\[
\frac{\Delta \sigma^{10}}{\Delta \varepsilon} = 10 \Delta \overline{\varepsilon}^{10}.
\]

The obvious correction is to plot \( \Delta \overline{\varepsilon}^i \); the total change in the secant modulus, along the ordinate. This is achieved by calculating:

\[
\Delta N^i = N(t^i) - N(t^{j^i}),
\]

where,

\( t^i \) is the time upon reaching the \( i \)th increment of constant strain,

\( t^{j^i} \) is the time just prior to ramping to the \( i + 1 \) increment of constant strain, and,

\( N(t^i) \) is the applied load (per unit width) at time, \( t \), during the \( i \)th increment of constant strain.

If one assumes that the thickness, \( m \), of the material remains constant, then dividing (2.6) through by \( m \) yields:
\[ \Delta \sigma' = \sigma(t'_1) - \sigma(t'_1), \quad (2.7) \]

where,

\[ \sigma(t') \] is the calculated laminate stress at time, \( t' \).

Unfortunately, the applied strain level varies slightly due to machine and control limitations. It is necessary, therefore, to normalize the recorded laminate stress found in (2.7) with the recorded strain, resulting in:

\[ \Delta \overline{\varepsilon} = \overline{\varepsilon}(t'_1) - \overline{\varepsilon}(t'_1), \quad (2.8) \]

where,

\[ \overline{\varepsilon}(t') = \frac{\sigma(t'_1)}{\varepsilon(t')} \quad , \quad (2.9) \]

and,

\[ \varepsilon(t') \] is the recorded strain at time, \( t' \).

\( \Delta \overline{\varepsilon}(t') \) is penned the "apparent modulus". It should not be confused with conventional engineering stiffness calculations. It is an instantaneous computation of the secant modulus (where both the load and strain are implicitly referred to the origin, though, in reality, upon unloading it is not likely that the response would pass through the origin). The "apparent modulus" is simply a mathematical manipulation of load, strain, and thickness, at time, \( t' \).

The second misrepresentation of the data found in [58,62] occurs when treating the data as continuous, i.e., not discrete. The data should be plotted as a histogram (see Figure 6) since discrete points were tested in the domain. The curve drawn in Figure 5 is a "smoothed" curve drawn to connect the data points. In this manner, it accentuates the "peaks" alluded to earlier, allowing the overall "trends" to be easily identified. Realistically, only when the testing increment, \( \Delta \varepsilon \), approaches zero, could such a continuous curve accurately represent the data. As it stands, the curve infers a continuous relationship between load drop and applied strain from the results of a limited number of discrete data points.
curve implies a continuous domain, all output in this work will show the discrete points simply connected by straight lines.

The conversion to "interrupted-ramp strain input" from a constant load input, and the graphical change in the output differentiate the test method employed in this study to the method used in [58,62]. This updated version will be known as the Incremental Strain Test. Clearly, these are progressive modifications of an original and insightful idea.

Test Procedure

The test procedure as described in [58] was completely manually-controlled. The physical demands on the tester as both controller and data-collector were inordinate. Computer-control was ideally suited for the simple, yet, time-consuming task. To this end, an IBM-PC XT was employed in conjunction with a 16 channel A/D, 2 channel D/A data acquisition system (DT2801-A) by Data Translation. Interactive software (PCLAB) was provided with the system, allowing the user to create his own PC-run program while incorporating various data acquisition commands. A screw terminal (DT752) by Data Translation permitted easy access to any/all electrical connections.

The test was run in strain control with all strains measured with a 1 in. extensometer. All test specimens were gripped in the machine according to the procedure delineated earlier, ensuring that both load and strain were simultaneously zeroed to avoid erroneous modulus measurements at low strains.

Voltage is output from the computer's D/A converter into the testing machine where it is then translated into a set strain level. The strain signal from the extensometer is output from the testing machine, converted to a digital signal by the A/D board and compared to the ideal value. The amount of discrepancy translates into a new voltage output, thus completing the closed-loop strain-control system. The load and strain output (as separate voltages) from the testing machine may be converted using the A/D board and stored in memory at any time.
Figure 6. Output from ILT represented as a histogram.
The actual test time and time loops are regulated with the PC’s internal clock. A schematic of this set-up is found in Figure 7 while a flowchart of the algorithm is found in Figure 8. The testing algorithm is briefly summarized below:

1. The specimen is ramped up to an initial strain level, $\varepsilon_0$.

2. Upon reaching this level, time, load, and strain are recorded and the secant modulus, $\overline{E}(\varepsilon_0)$ is calculated according to (2.9).

3. At chosen time intervals, $\Delta t$, the above data are again recorded.

4. This strain is held constant until the chosen time loop, $t_f - t_0$, is completed.

5. Strain is ramped up to its new level, according to (2.2). The increment number is now $i = 1$.

6. The whole process then continues until the strain reaches a chosen termination strain, $\varepsilon_n$, or failure occurs.

7. At this point, the strain is ramped down to where the load registers less than 60 lb., and the test is ended.

The data (time, strain, load, and $A$ modulus) may be recalled and processed at any time. The algorithm readily admits to the use of several non-destructive techniques (edge replication, X-ray radiography, and acoustic emission, in particular).

The generality of the program allows the user to introduce variability into the test. The investigator may choose:

1. the starting strain level, $\varepsilon_0$,

2. the strain increment, $\Delta \varepsilon$,

3. the time between sampling data, $\Delta t$. 

Experimental Investigation
Figure 7. Schematic of the IST computer-control system.
Figure 8. Flowchart of the IST algorithm.
4. the time between ramping strain (the hold time), $t_r - t_h$.

5. the termination strain, $\varepsilon_n$.

6. the strain level(s) at which $\Delta \varepsilon$ changes, and,

7. the new strain increment(s), $\Delta \varepsilon'$.

Altering any combination of these variables creates completely new test. A complete program listing is found in Appendix A.

**Material and Test Schematic**

It would be quite prudent at this time to review, in tabular form, the material tested and the mechanical tests performed (see Table 1).

**Damage Evaluation**

Damage evaluation is an essential and invaluable step toward the understanding of the physical response of composite laminates. Non-destructive evaluation techniques interrogate the material's integrity, revealing the type, location, extent, degree, chronology, and role that damage plays in the laminate's performance. A review of the techniques employed in this study is included below.
Table 1. Material and Test Schematic.

<table>
<thead>
<tr>
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</thead>
<tbody>
<tr>
<td></td>
<td>B</td>
<td>I</td>
<td>B</td>
</tr>
</tbody>
</table>

Tests

- Monotonic Tension:  
  - X  X  X  X  
- Monotonic Compression:  
  - X  X  |
- T-C Fatigue:  
  - X  X  |
- Incremental Strain:  
  - X  X  

Experimental Investigation
Non-destructive Evaluation

Ultrasonic C-scan

An ultrasonic C-scan was performed on each as-received panel prior to machining as a test of quality assurance. In this method, a transducer emits bursts of high-frequency sonic waves through a couplant (in this case, water) to the panel. The wave will be reflected, refracted, and transmitted at each interface it encounters. In the "pulse-echo" mode of operation (exclusively used here) a sensor measures the energy of the reflected waves.

A gate is set on the amplitude of the energy that is reflected through the "acceptable" material. Any voids or defects in the material will produce superfluous interfaces, attenuating the energy of the reflected signal. This attenuation will not pass through the gate and will reproduce as an anomaly. From Figure 9, the black represents "acceptable" material while the white areas pinpoint the defects. In this manner, ultrasonic C-scan can determine the location, extent (in two-dimensions) and degree of damage. It becomes a totally subjective decision whether to accept a panel for testing. An excellent description of the test method and test equipment used in this study is found in [63].

Penetrant-Enhanced X-ray Radiography

Penetrant enhanced X-ray radiography is clearly one of the most effective non-destructive techniques used to evaluate the damage (or lack of) in a composite laminate. The procedure required to produce these radiographs is quick and easy, adding to its appeal. The method is detailed below.

Most damage in an unnotched coupon initiates from the edge for several reasons. These reasons may include edge flaws due to machining, and/or stress concentrations due to the...
Figure 9. Output from an ultrasonic C-scan of a composite panel.
edge effect with regard to mechanical, thermal, or moisture loading. In any case, when damage intersects a free surface it provides a path for a penetrant. The penetrant must have at least these two qualities to be effective:

1. opacity to X-rays, in order to provide the necessary contrast in the radiograph, and,
2. a high surface tension-to-density ratio, to promote capillary action in the defect.

In addition, the penetrant, ideally, should be non-corrosive to the matrix material.

Experience has shown that a solution of zinc iodide adequately satisfies this criteria. The solution (60 g of ZnI₂ mixed in 10 ml each of water, isopropyl alcohol, and Kodak Photo-Flo 200) is most effectively applied while the specimen is under load since this tends to "open" most damage, ensuring optimum seepage. It is important to allow a sufficient time to elapse after application (24 hours or more); convincing evidence shows that impatience often produces misleading radiographs [62]. The excess penetrant is removed from all free surfaces with acetone.

The X-ray cabinet utilized (Hewlett Packard 4380SN Faxitron Series X-ray System) introduces three variables into each radiograph; the distance from the emitting element, the applied voltage, and the exposure time. These variables are all juggled with the intent of producing the best radiograph for that specimen. All three variables are strongly dependent on laminate thickness. Single emulsion (Kodak SR-5) or double emulsion (Kodak M-5) X-ray film was used interchangeably, having found no significant difference in the resulting radiograph. Exposure time and applied voltage will change markedly (by more than a factor of 2) when going from double to single emulsion film.

A summary of each material and the X-ray settings used — based on double emulsion film — is found in Table 2.
Table 2. X-Ray Settings for each Material Type.

<table>
<thead>
<tr>
<th>Material</th>
<th>Thick. (in.)</th>
<th>Distance (in.)</th>
<th>Voltage (kVp)</th>
<th>Time (s)</th>
</tr>
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<tbody>
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<td></td>
<td></td>
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<td></td>
</tr>
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<td>B</td>
<td>0.0915</td>
<td>16</td>
<td>30</td>
<td>30</td>
</tr>
<tr>
<td>I</td>
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<td>16</td>
<td>30</td>
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<td>70</td>
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</tr>
<tr>
<td>B</td>
<td>0.0390</td>
<td>16</td>
<td>35</td>
<td>15</td>
</tr>
</tbody>
</table>
Acoustic Emission Monitoring

Acoustic Emission is defined as the high-frequency stress waves generated by the rapid release of strain energy that occurs within a material during crack growth, plastic deformation or phase transformation [64].

This definition highlights the appeal that AE monitoring has to composite researchers. Any technique capable of detecting and quantifying the release of strain energy would be invaluable to the study of composites. When damage is produced in a material it is always accompanied by a release of stored strain energy. Monitoring AE while loading a composite laminate reveals the chronology of damage events. With an effective quantifying technique and experienced interpretation skills, one can distinguish separate damage modes.

Acoustic Emission monitoring was employed as a non-destructive technique during the Incremental Strain Test (IST) with the intention of comparing their respective results. Any correlation that might exist between the results would lend considerable validation to the IST. In particular, the stiffness relaxation (increased compliance) peaks displayed from the IST were hoped to correlate directly in time (at similar levels of applied strain) to an increase in AE activity.

To this end, an AE transducer was attached to a test specimen according to Figure 10. It was necessary to use high-vacuum silicone grease as a couplant since conventional couplants were found to degrade and evaporate over the 12+ hour tests. Care was taken to ensure that the contact pressure prevented time-dependent slipping of the transducer. A schematic of the signal processing set-up is shown in Figure 11. The signal from the transducer was fed into a pre-amplifier. The output from pre-amp (at 60 dB) was sent through an oscilloscope (for visual examination) and then to an RMS meter. The analog RMS voltage output from the RMS meter was sampled and stored by the A/D board on the IBM PC at a rate of approximately 3 Hz. The settings on each piece of equipment were selected to enhance the signal monitoring. Upon achieving a representative signal, the settings remained untouched for the duration of a single test. Extreme care was not taken in reproducing the exact settings for each test. The goal was to correlate AE and IST results, not AE results from different tests.

Experimental Investigation
The algorithm for measuring AE during the IST is found below:

1. A line in the computer program (Appendix A) calls for the board to sample the RMS meter through an A/D conversion.

2. The digital resultant is added to the previous number stored in the AE output variable (which is zero, if it is the first sample).

3. The RMS meter is sampled at a rate dictated by the time required for the program to return to the sampling command, which proved to be approximately 3 times a second.

4. The additive result of the AE variable is squared in order to reflect a "power" term. This number is output (and stored) when the time interval for data sampling has elapsed.

5. The AE variable is reset to zero and the process continued.

A separate program (Appendix A) was written to manipulate the AE data obtained from the test. The data displayed for each constant strain level is scanned to determine the minimum value. This value is assumed to be a measure of equilibrium "noise" for the applied strain. This "noise" value is subtracted from each data point (for that level of strain) creating a number representing an amount of activity above equilibrium. Then each of these new numbers are added together to measure a cumulative amount of activity during the time when the particular strain level was held constant. Therefore, one number quantifies the total AE activity per each applied strain increment.

**Stiffness Monitoring**

Monitoring the change of the static secant stiffness modulus during the life of a fatigue-cycled composite laminate has been shown by many investigators to be an exceptional measure of progressive damage development [65-69]. The unnotched laminate stiffness re-
Figure 10. AE transducer attached to an iST specimen (not to scale).
Figure 11. Schematic of the AE signal processing.
ductions due to matrix cracking [2] and delamination [1] have been studied experimentally and analytically, and are generally well-understood. In fatigue loading, where the dominant damage modes are, usually, matrix cracking and delamination, it is not surprising that stiffness reduction correlates well with damage progression. For this reason, stiffness monitoring was performed during the fatigue loading of the AS4/C1808 laminates. The data from such monitoring is commonly displayed as a plot of normalized stiffness \( \frac{E}{E_0} \) vs. normalized cycles to failure \( \frac{N}{N_f} \) (see Figure 12). The plots obtained from the B and I laminates' cyclic response will provide a valuable comparative measure of damage development.

The procedure is very straightforward. Before fatigue cycling begins, the specimen's strain (measured with an extensometer in this study) is zeroed at zero load. The specimen is then slowly ramped to the maximum amplitude of its fatigue-load excursion. The load and strain are recorded at this load level. The resultant stiffness calculated is termed the "secant" tensile modulus since it is a measure of the slope of the line running from the origin to the measured point (in stress-strain space). The compressive secant modulus is calculated from the strain measured at the minimum load excursion. The difference in the secant modulus and tangent modulus becomes apparent only when the stress-strain response is non-linear. At regular intervals during the test (every \( x \) number of cycles or, similarly, every \( y \) minutes) the cycling is interrupted and the load is returned to zero. The residual strain is recorded, then zeroed. The tensile and compressive secant moduli are measured using the above procedure and may be normalized to their original respective modulus. Having recorded the number of cycles accumulated, the test is resumed. If continued until failure, the data (in terms of applied cycles) may be normalized to the number of cycles to failure. This subtle maneuver enables one to make a direct comparison between long-term and short-term responses.
Figure 12. Plot of normalized stiffness vs. normalized life for an AS4/C1808 specimen cycled at a low-load amplitude.
**Edge Replication**

Prior to the advent of the scanning electron microscope (SEM), the transmission electron microscope (TEM) was the superior tool for fractography studies of metals. It was necessary, however, to make a replica of the surface detail in order to allow the through-transmission of the electron beam. A popular technique, used for both cleaning the specimen and for replication, employed cellulose acetate as the replicating medium. When applied to the desired surface and wetted with acetone, the acetate would partially melt into the surface. Upon drying, the acetate was peeled away, carrying with it a topographic impression of the fracture surface. Though this replica demanded further preparation before observation in the TEM, the replication technique was fast, simple, and non-destructive [70].

Several years ago, this replicating technique was applied to the study of edge damage in composite laminates [71-73]. The laminate free-edge, highly susceptible to damage in the form of off-axis matrix cracking and delamination, is an ideal site for microscopic study through surface replication. An essential step prior to such a study involves surface preparation. Since common laboratory specimens are sized by machining through-the-thickness, it is necessary to polish the edge. A highly polished edge reveals any inherent material defects and isolates subsequent material failures. Thus, a “clean", informative replica may be obtained. The procedure used in this study for creating an edge replica is thoroughly detailed in [63].
Experimental Results

The contents of this chapter must address — at the very least — this question: Does the presence of the interlayer alter the laminate’s response to its loading condition? The answer to this will simply involve the quantitative comparison of experimental results. This chapter will be partitioned into three sections corresponding to the three loading conditions; monotonic loading, fatigue loading, and incremental strain.

Monotonic Loading

AS4/C985

The monotonic tensile data for this material with and without interlayers are located in Table 3. Compressive data were not obtained since the laminates were too thin to obtain failure without gross buckling or equipment damage.
Table 3. AS4/C985 Monotonic Tensile Data.

<table>
<thead>
<tr>
<th>Spec.</th>
<th>Area (in²)</th>
<th>Fail. Load (ksi)</th>
<th>Fail. Strain (με)</th>
<th>Strength (ksi)</th>
<th>E₂ (Msi)</th>
</tr>
</thead>
<tbody>
<tr>
<td>B-1</td>
<td>0.0919</td>
<td>8.14</td>
<td>13500</td>
<td>88.6</td>
<td>6.67</td>
</tr>
<tr>
<td>B-2</td>
<td>0.0914</td>
<td>8.70</td>
<td>14300</td>
<td>95.2</td>
<td>6.88</td>
</tr>
<tr>
<td>B-9</td>
<td>0.0918</td>
<td>8.06</td>
<td>13500</td>
<td>87.8</td>
<td>6.63</td>
</tr>
<tr>
<td>Avg.</td>
<td>0.0917</td>
<td>8.30</td>
<td>13770</td>
<td>90.5</td>
<td>6.73</td>
</tr>
</tbody>
</table>

Interlayered

<table>
<thead>
<tr>
<th>Spec.</th>
<th>Area (in²)</th>
<th>Fail. Load (ksi)</th>
<th>Fail. Strain (με)</th>
<th>Strength (ksi)</th>
<th>E₂ (Msi)</th>
</tr>
</thead>
<tbody>
<tr>
<td>I-1</td>
<td>0.0947</td>
<td>8.80</td>
<td>14800</td>
<td>93.0</td>
<td>6.62</td>
</tr>
<tr>
<td>I-2</td>
<td>0.0978</td>
<td>9.03</td>
<td>14700</td>
<td>92.5</td>
<td>6.47</td>
</tr>
<tr>
<td>I-9</td>
<td>0.0974</td>
<td>9.06</td>
<td>15300</td>
<td>93.0</td>
<td>6.07</td>
</tr>
<tr>
<td>Avg.</td>
<td>0.0966</td>
<td>8.96</td>
<td>14930</td>
<td>92.8</td>
<td>6.39</td>
</tr>
</tbody>
</table>
The most striking comparative result between the I and B laminates is the difference in load and strain at failure. On average, the load and strain at failure for the I laminate is 8% higher than the B laminate. This distinction is aided by the fact that the B specimen (B-2) with the highest load at failure is still below the I specimen (I-1) with the lowest load to failure. It is interesting to note the minimal deviation in the load and strain at failure for the I laminate compared to the B laminate. As a general comment, the results of the baseline material, a 16-ply, quasi-isotropic, AS4/C985, are, indeed, impressive. An average failure strain of over 1.35% for this lay-up is quite remarkable when one considers that AS4 fiber, alone, is quoted as having a failure strain of 1.5% [5]. With the government hoping to incorporate design strains of 0.6%, a system such as this — in a quasi-isotropic lay-up — possesses a safety factor greater than two.

Comparing the laminate stiffness and strength of these two laminate types can be misleading, however. The interlayer adds, on average (of all AS4/C985 specimens tested), 6% to the laminate thickness, which translates into the same increase in cross-sectional area for 1 in. wide coupons. Strength is then computed by dividing the failure load with the cross-sectional area. This is an appropriate comparison using identical materials with identical lay-ups. The interlayer, however, is a low-modulus (on the order of 0.5 Msi or less [74]) layer, approximately 1/10 the thickness of the constituent lamina; for this reason, it carries little to no load in the laminate. It is better, perhaps, to approximate the load carried by the interlayer prior to failure, subtract it from the total load carried, and then treat the I and B laminates as having equal thickness, in order to calculate comparable strengths.

The moduli, $E'$, listed in Table 3, are subjective results. The number shown is the value calculated from the plot of load vs. strain obtained from an $X-Y$ plotter. The slope of the early linear portion was divided by the cross-sectional area of the specimen for the result. Though the precise stiffness may be suspect due to inherent non-linearity in the plot, the trend of lower stiffnesses in the I laminates compared to the B laminates is to be expected. Adding several (thin) layers of low-modulus material to a laminate will cause its effective laminate stiffness, $\bar{E}$, to decrease.
A good approximation of this reduction may be obtained using classical lamination theory (CLT). If \( t_i \) is the thickness of the I laminate and \( t_b \) is the thickness of the B laminate, then the ratio of the effective laminate stiffness of the I laminate, \( \overline{E}_i \), to the effective laminate stiffness of the B laminate, \( \overline{E}_b \), for a quasi-isotropic laminate is:

\[
\frac{\overline{E}_i}{\overline{E}_b} \approx \frac{1 + \frac{U'_1}{U_1} r}{1 + r},
\]

(3.1)

where, \( U'_1 \) and \( U_1 \) are the first invariants of the ply moduli [75] of the isotropic interlayer material and the baseline composite lamina, respectively, and (see Appendix B for details),

\[
r = \frac{t_i - t_b}{t_b}.
\]

(3.2)

With these results one can estimate, using CLT, the load carried by the interlayers in the laminate. Assuming no ply failures:

\[
\frac{N_i}{N_b} = \frac{\overline{E}_i}{\overline{E}_b} \frac{t_i}{t_b} \frac{\varepsilon_i}{\varepsilon_b}.
\]

(3.3)

At equal applied strains, \( \varepsilon_i = \varepsilon_b \), from (3.1)-(3.3):

\[
\frac{N_i}{N_b} = \frac{\overline{E}_i}{\overline{E}_b} (1 + r).
\]

(3.4)

From Table 3, on average, \( \frac{\overline{E}_i}{\overline{E}_b} = 0.949 \) and \( r = 0.0636 \), so:

\[
\frac{N_i}{N_b} = (0.949)(1.0636) = 1.0099.
\]

(3.5)

Therefore, the I laminate, due to the presence of the interlayer, carries 1% more load than the B laminate at equal applied strains (assuming no ply failure). An estimation of the interlayer's share of the failure load is desired, however. To this end, Specimens I-14 and B-13
were ramped quickly to their near failure strain (14500 με and 13500 με, respectively) and the secant modulus of each recorded. In this case, \( \frac{\bar{E}}{E_0} = 0.950 \), so, again, \( \frac{N_i}{N_o} \cong 1.01 \). It may be argued that this analysis is not appropriate since ply damage has occurred near failure.

By approximating the load carried by the interlayers near failure, the "strength" of the I laminate can be adjusted, as delineated above. On average, the "adjusted strength," \( \bar{X}' \), (from Table 3) becomes:

\[
\bar{X}' = \frac{(8.96 - (8.96)(0.01)) \text{ kips}}{0.0917 \text{ in}^2} = 96.7 \text{ ksi}.
\]

This value better reflects the increase in strength due to the interlayer's presence. Despite subtracting the 1% additional load carried by interlayers, the I laminate still reveals a 7% increase in load at failure over the B laminate.

**AS4/C1808**

**Tensile Data**

The monotonic tensile data for this material with and without interlayers are located in Table 4. Compressive results were obtained for this laminate due to their large cross-sectional areas. It should be noted that all specimens were notched with a 3/8 in. center-hole. Strength and stiffness were calculated using the gross cross-sectional area. Strain was measured with an extensometer mounted across the hole.

Again, the most noticeable difference occurs in ultimate load. On average, the I laminate boasts a 13% increase in this value over the B laminate. The I laminate shows an average 5% increase in ultimate strain over the B laminate, yet, this is suspect when considering the variability in the B data.
Table 4. AS4/C1808 Monotonic Tensile Data.

<table>
<thead>
<tr>
<th>Spec.</th>
<th>Area (in²)</th>
<th>Fail. Load (ksi)</th>
<th>Fail. Strain (µε)</th>
<th>Strength (ksi)</th>
<th>$E^e$ (Msi)</th>
</tr>
</thead>
<tbody>
<tr>
<td>A-2-14</td>
<td>0.273</td>
<td>10.50</td>
<td>-</td>
<td>38.5</td>
<td>-</td>
</tr>
<tr>
<td>A-1-12</td>
<td>0.268</td>
<td>10.70</td>
<td>9900</td>
<td>39.9</td>
<td>4.40</td>
</tr>
<tr>
<td>A-1-10</td>
<td>0.268</td>
<td>10.70</td>
<td>8800</td>
<td>39.9</td>
<td>4.65</td>
</tr>
<tr>
<td>Avg.</td>
<td>0.270</td>
<td>10.60</td>
<td>9350</td>
<td>39.4</td>
<td>4.53</td>
</tr>
</tbody>
</table>

<table>
<thead>
<tr>
<th>Spec.</th>
<th>Area (in²)</th>
<th>Fail. Load (ksi)</th>
<th>Fail. Strain (µε)</th>
<th>Strength (ksi)</th>
<th>$E^e$ (Msi)</th>
</tr>
</thead>
<tbody>
<tr>
<td>C-1</td>
<td>0.2970</td>
<td>11.69</td>
<td>9730</td>
<td>39.4</td>
<td>3.86</td>
</tr>
<tr>
<td>C-2</td>
<td>0.2947</td>
<td>12.23</td>
<td>10000</td>
<td>41.5</td>
<td>4.35</td>
</tr>
<tr>
<td>Avg.</td>
<td>0.2959</td>
<td>11.96</td>
<td>9870</td>
<td>40.5</td>
<td>4.11</td>
</tr>
</tbody>
</table>
The difference in laminate strength seems insignificant, yet, the same arguments apply here as did for the AS4/C985 laminates. Since the lamina in the undamaged I and B laminates carry the same load at the same constant applied strain (the only difference is due to a minute disparity between laminate Poisson ratio, $\nu_{xy}$), the interlayers produce the additional load-carrying capability of the I laminate. The amount of load carried by the interlayers is not readily approximated from (3.4) since the measured values of stiffness are suspect. Equation (3.1) provides no help in estimating $\frac{E_i}{E_n}$ since $U_i$ is unknown for AS4/C1808 [74]. Despite this, the arguments presented should convince one that the “adjusted strength” of the I laminate should show an increase comparable to the 13% increase in ultimate load over the B laminate.

The presence of the interlayer accounted for an 8% increase in ultimate tensile load over the baseline AS4/C985 and a 13% increase in ultimate tensile load over the baseline AS4/C1808. The trend in higher ultimate loads due to the interlayer’s presence is indisputable. Concurrence with this trend is found in the results of [19-21,34].

**Compressive Data**

The monotonic compressive data for this material with and without interlayers is located in Table 5.

The ultimate performance of the I laminates reveals no significant change when the loading changes from tensile to compressive; all of the measured properties stay, basically, the same. The ultimate performance of the B laminates, however, changes radically. The compressive ultimate load is, on average, 16% higher, while the ultimate strain is 49% higher than the tensile ultimate results. There is no obvious explanation for this surge in compressive performance [76]. If this is typical of AS4/C1808 laminates in this lay-up configuration, then, one should note that the interlayered laminate does not exhibit this trend.
Table 5. AS4/C1808 Monotonic Compressive Data.

Baseline

<table>
<thead>
<tr>
<th>Spec.</th>
<th>Area (in²)</th>
<th>Fail. Load (ksi)</th>
<th>Fail. Strain (µε)</th>
<th>Strength (ksi)</th>
<th>$E_\sigma$ (Msi)</th>
</tr>
</thead>
<tbody>
<tr>
<td>A-1-9</td>
<td>0.271</td>
<td>12.50</td>
<td>14600</td>
<td>46.1</td>
<td>4.50</td>
</tr>
<tr>
<td>A-1-11</td>
<td>0.272</td>
<td>12.00</td>
<td>13300</td>
<td>44.1</td>
<td>4.50</td>
</tr>
<tr>
<td>Avg.</td>
<td>0.2715</td>
<td>12.25</td>
<td>13950</td>
<td>45.1</td>
<td>4.50</td>
</tr>
</tbody>
</table>

Interlayered

<table>
<thead>
<tr>
<th>Spec.</th>
<th>Area (in²)</th>
<th>Fail. Load (ksi)</th>
<th>Fail. Strain (µε)</th>
<th>Strength (ksi)</th>
<th>$E_\sigma$ (Msi)</th>
</tr>
</thead>
<tbody>
<tr>
<td>C-3</td>
<td>0.2991</td>
<td>11.56</td>
<td>&gt;10000</td>
<td>38.8</td>
<td>4.02</td>
</tr>
<tr>
<td>C-4</td>
<td>0.2976</td>
<td>12.46</td>
<td>&gt;10000</td>
<td>41.9</td>
<td>4.05</td>
</tr>
<tr>
<td>Avg.</td>
<td>0.2984</td>
<td>12.01</td>
<td>&gt;10000</td>
<td>40.3</td>
<td>4.03</td>
</tr>
</tbody>
</table>
As mentioned earlier, the 8-ply, quasi-isotropic, AS4/3502 material served as a baseline, first-generation, brittle-epoxy resin system. Its response to various loading conditions is well-documented and well-known. For this reason, the AS4/3502 played the "guinea pig" for the initial and validation phases in the development of the Incremental Strain Test (IST). The monotonic tensile results for the coupons manufactured at VPI are located in Table 6. As with the AS4/C985, monotonic compressive data were not obtainable due to insufficient laminate thickness.

**Fatigue Loading**

**Notched AS4/C1808**

Center-notched I and B laminates comprised of 32 ply, quasi-isotropic, AS4/C1808 were subjected to fully-reversed (R = -1), tension-compression fatigue cycling. As earlier mentioned, the I material was to be tested in a manner which would allow for comparison to a baseline test matrix already under way. The B test matrix (performed by C. E. Bakis) focused on differentiating between high-load — low-load notched response. The focus of the present study is to assess the effect interlayering has on the laminate response. With limited material, the approach undertaken was to "nestle" in between the two load extremes. By choosing a "middle" load level, it was reasoned, a response indicative of that level, i.e., falling in between the response of the high-low extremes, would suggest that the interlayer's presence had little effect. Yet, if the response seemed anomalous to its particular load level, i.e., tended towards
Table 6. AS4/3502 Monotonic Tensile Data.

<table>
<thead>
<tr>
<th>Spec.</th>
<th>Area (in²)</th>
<th>Fail. Load (ksi)</th>
<th>Fail. Strain (με)</th>
<th>Strength (ksi)</th>
<th>$E_0$ (Msi)</th>
</tr>
</thead>
<tbody>
<tr>
<td>S-1</td>
<td>0.0399</td>
<td>3.30</td>
<td>12500</td>
<td>82.7</td>
<td>7.39</td>
</tr>
<tr>
<td>S-2</td>
<td>0.0389</td>
<td>2.85</td>
<td>10200</td>
<td>73.2</td>
<td>7.74</td>
</tr>
<tr>
<td>S-20</td>
<td>0.0389</td>
<td>3.78</td>
<td>11400</td>
<td>97.2</td>
<td>8.86</td>
</tr>
<tr>
<td>Avg.</td>
<td>0.0392</td>
<td>3.31</td>
<td>11370</td>
<td>84.4</td>
<td>8.00</td>
</tr>
</tbody>
</table>
one of the extremes, the effect of the interlayer would be quite visible. Quantitative response was measured in the laminates by stiffness monitoring and residual strength testing. The results of these tests are found below.

**Stiffness Monitoring**

A comparison of the normalized stiffness vs. normalized life curves corresponding to the three applied load levels is found in Figure 13. The high-load and low-load response of the B material is the upper and lower curve, respectively. The low-load response is highlighted by the three large stiffness reductions. These reductions correlate to the growth of large-scale delaminations. The overall stiffness reduction indicates the preponderance of matrix damage in the low-load response. The high-load curve reveals a steady degradation in stiffness until near failure, where the stiffness decreases rapidly. This indicates that the matrix damage is not as extensive as the low-load case. The curve corresponding to the I response nestles between the two extremes. The slope of the curve is quite similar to the high-load response of the B material through 90% of normalized life. The first 5% and final 5% of the normalized life is similar, instead, to the low-load response — exhibiting comparable wear-in and wear-out. Therefore, one might conclude from the comparative stiffness curves that the response of the I material falls naturally between the two load level extremes of the B material.

**Residual Strength Testing**

Intensive work has been performed with the hopes of predicting the residual notched strength given the fraction of remaining fatigue life [77]. Matrix damage around center-notches tends to reduce the stress concentrations inherent in the geometry. This gives rise to the well-documented instance of an increase in notched residual strength upon the introduction of damage. Recently, in a work by Simonds, et al. [43], the life of a center-notched
Figure 13. Plot of normalized stiffness vs. normalized life for three AS4/C1806 laminates under different applied load amplitudes.
lamine containing a tough matrix system was shown to be strongly dependent on the applied load amplitude. Low-load levels produced extensive delaminations and matrix cracking and resulted in long lives with good tensile strength retention, while high-load levels produced highly localized damage in the form of fiber fracture, matrix cracks, and local delaminations, resulting in short lives with poor tensile strength retention. It was hypothesized that the energy channeled into the considerable matrix damage (found at low loads) prevents it from being focused on fiber fracture — a true strength-limiting mechanism. At high loads the fracture energy is focused on fiber fracture. Hence, it may be reasoned, extensive matrix damage — while it severely degrades the stiffness and debilitates compressive strength — may aid considerably in tensile strength retention of notched laminates. With this in mind, one should examine the results of the comparative residual tensile strength testing located in Table 7.

Residual tensile strength was found at "early" (15-20% of life), "middle" (50-70% of life), and "late" (>90% of life), for the three load levels. All three showed an increase in residual tensile strength for "early" life. This could be attributed to the damage originating at the edge of the hole that tends to reduce the stress concentration. The residual tensile strength of the I material contrasts sharply with the results from the B material at "middle" life. The B material — at both high and low load levels — reveals a further increase in residual strength, likely owing to the further relaxation in stress concentrations due to matrix damage, and, in particular, the presence of global delaminations. Applying earlier reasoning, if these delaminations were suppressed or averted by the toughened interlayers, fiber fracture may ensue, reducing the residual strength. The results in Table 7 provide some support to this claim.

At "late" life, the residual tensile strengths vastly differ. The low-load B material still shows no tensile strength reduction. The failure of this laminate at the advanced state of damage is always a compression-instability failure. No new mechanism is introduced that would reduce the residual tensile strength. The residual tensile strength of the B material under high-load fatigue at "late" life exhibits significant reduction. It may be conjectured that a transition from matrix failure to fiber failure occurs toward the end of life, reducing the tensile strength. The I material has no appreciable change in tensile strength at "late" life.
Table 7. Residual Strength of Fatigue Loaded AS4/C1808.

<table>
<thead>
<tr>
<th>% of life</th>
<th>l — mid.</th>
<th>B — high</th>
<th>B — low</th>
</tr>
</thead>
<tbody>
<tr>
<td>15-20</td>
<td>1.14</td>
<td>1.11</td>
<td>1.10</td>
</tr>
<tr>
<td>50-70</td>
<td>0.97</td>
<td>1.19</td>
<td>1.21</td>
</tr>
<tr>
<td>90+</td>
<td>1.03</td>
<td>0.88</td>
<td>1.17</td>
</tr>
</tbody>
</table>

Each number represents the ratio of the residual tensile strength to the monotonic tensile strength.
Perhaps the damage producing the strength reduction at "middle" life stabilizes and subsequent damage manifests as delaminations, thus producing little change in the tensile strength. This laminate, however, does fail in tension.

**Incremental Strain Loading**

**Interpretation of Results**

The IST results performed on a specimen (A-10) from the AS4/3502 laminate are shown in Figure 14. The highlights of this display include the following:

1. Results obtained from small magnitudes of applied strain (less than 3000 με) are suspect and anomalous. This region, penned "undamage" in [58], has no physical basis, and, therefore, appears to be caused by system deficiencies. At such low strains, the controlling voltage may easily be absorbed by system noise. To avoid this, tests on all materials are often begun at strains above 3000 με but below the onset of ply damage. Even at these higher initial strains, the beginning of a test always yields suspect results. Hypothetically, this might be caused by a "wear-in" period, i.e., the extensometer is situating in its tabs, the machine is responding to its internal deformations, etc.

2. "Peaks" in the curve correspond to off-axis ply damage. Results from this study and those in [58,62] correlate the damage state recorded by X-ray radiography to a drop in the applied stress on the IST curve. In Figure 14, at 4000 με the laminate is becoming more compliant, yet, the strain level is below those levels causing gross changes in stiffness (5500-7000 με). A radiograph of a specimen (A-5) at 4000 με is shown in Figure 15. No
Figure 14. IST output from Specimen A-10.

Experimental Results
90° ply cracks are found to have spanned the coupon. The damage present is likely due to machining, yet, has probably experienced some growth caused by the applied strain. A radiograph of a specimen (A-12) loaded to 6500 με is shown in Figure 16. The coupon is replete with 90° ply damage and is accompanied by some -45° ply cracking. The damage present correlates exceptionally well with the magnitude of stiffness degradation found in Figure 14.

3. Stiffness degradation in regions prior to the occurrence or growth of matrix damage (say, 3000-5000 με in Figure 14) may be attributed, at least in part, to viscoelastic laminate behavior. If either the strain "zero" or the actual extensometer drifts, then these quantitative results are suspect. Otherwise, while maintaining constant strain (deformation), if the compliance of an off-axis ply increases due to viscoelastic relaxation, the stress (load) carried in that ply will reduce. Therefore, at constant strain, viscoelastic behavior may result in the reduction in laminate stiffness. At advanced levels of strain, the effects of viscoelastic behavior, if present, are nearly impossible to distinguish from the effects of matrix damage.

If the early behavior is predominately viscoelastic creep, then, by definition, the effect is reversible. The IST — by maintaining a constant and ever-increasing strain — does not allow for any reversibility. It is unclear from the IST results whether the viscoelastic response is linear. The viscoelastic behavior of each laminate type could be effectively characterized by independent creep and/or relaxation tests.

4. Durability earlier was defined as a measure of a material's resistance to stiffness loss. The area under the IST curve at a given strain level (when considering the IST as a histogram) provides a comparative measure of durability in response to time-dependent tensile loading.
Figure 15. Radiograph of Specimen A-5 at 4000 $\mu s$ under IST loading.
Figure 16. Radiograph of Specimen A-12 at 6500 μs under IST loading.
5. Damage tolerance was earlier defined as a measure of a material’s resistance to failure when damage is present. The material’s strain at failure in the IST provides a comparative measure of damage tolerance in response to time-dependent tensile loading.

The information, as listed above, could prove quite valuable towards characterizing composite response. An attempt at “validating” the results should be made. To this end, coupons of each laminate type were tested using the IST, with hopes that “validation” could be achieved through:

1. Correlation — results of the IST were compared to the results obtained from acoustic emission (AE) monitoring.

2. Reproduction — obtaining like results from two separate coupons of identical material under like loading conditions.

A discussion of each method follows.

**Correlation**

As detailed earlier, acoustic emission (AE) activity was quantified and processed to reflect an amount of “energy released” per each strain increment. These values were plotted concurrently with the change in apparent modulus during an IST of an AS4/3502 specimen (A-12) in Figure 17. The correlation in the two curves is stunning. Referring to Figure 14, in AS4/3502 a “peak” is typically seen between 5500-7000 με. With this in mind, an IST was run between 3000 με and 6500 με, hoping to capture this activity. The AE activity as plotted substantiates the claim that ply damage accounts for the “peaks” in the data. Since AE is a measure of strain energy release — a release due to crack growth and other deformation mechanisms — it should correlate favorably to a change in compliance, since this change occurs due to the same mechanisms. A similar test was run using an I specimen (I-11) of AS4/C985 (see Figure 18). Again, the correlation between the two tests is exceptional. From Figures 17 and 18, one
may conclude that the IST is quite successful in isolating and highlighting the strain regimes in which ply damage occurs.

**Reproduction**

A valid test procedure must have the necessary condition that its results, under like conditions, are reproducible. The lack of reproducibility indicates the presence of unaccounted test variables and, more importantly, prevents the measurement of a physical quantity.

In order that the results of the IST be credible, coupons of the same material (cut from the same panel) should yield, under like test variables, similar results. Material dissimilarity would likely ensure dissimilar results. For this reason, reproducibility tests were conducted using the AS4/C985 material rather than the AS4/3502. The latter material was processed with poor quality control and is quite susceptible to damage caused by machining. Uniformity in the virgin coupons is a necessity. The comparative results of two specimens, B-10 and B-11, are shown in Figure 19. If one ignores the initial disparity, the two curves are quite comparable. Each curve exhibits a "double peak" from $9500 \mu e$ to $12500 \mu e$, unique to this laminate. The drop in apparent modulus in this strain domain is strikingly similar. These results indicate that a satisfactory level of reproducibility may be met, thus lending credibility to the test procedure.

**AS4/C985**

The IST results of a "long-life" (here, "long-life" refers to the final strain increment attained with respect to its static ultimate strain) B specimen (B-11) is shown in Figure 20. The essential observations from this display include:
Figure 17. IST output vs. AE output from Specimen A-12.
Figure 18. IST output vs. AE output from Specimen I-11.
Figure 19. IST output from Specimen B-10 vs. Specimen B-11.
Figure 20. IST output from Specimen B-11.

Experimental Results
1. This particular test was begun at 3000 $\mu e$ in hopes that the "undamage" portion seen in Figure 14 could be avoided. This was not the case, however. The first four data points indicate the presence of either control system deficiencies or other unaccounted variables. In a further attempt to avoid this non-physical region, later tests were begun at an initial strain of 5000 $\mu e$ or higher.

2. Two sets of "double peaks" (5500-8000 $\mu e$ and 9500-12500 $\mu e$) appear with surprising uniformity. One might postulate that the first double peak corresponds to matrix viscoelastic response. X-ray radiography indicates, however, that the second double peak corresponds to the progressive failure of the 90° plies. Therefore, in Specimen B-11, 90° ply failure (or, FPF) is seen to begin at 9500-10000 $\mu e$. This is a distinctive result since, as alluded to earlier, new design limits are striving toward 8000 $\mu e$ maximum strain loading.

3. There is no apparent reason for the second peak in the "double peak" occurring from 10500 $\mu e$ to 12500 $\mu e$. One might guess that the -45° plies begin to fail then. A radiograph taken of a specimen (B-7) at 12000 $\mu e$ (Figure 21) shows some -45° cracking, yet, no cracks span the width of the specimen. Another specimen (B-10) at 13500 $\mu e$ (Figure 22) shows a similar damage pattern to the one captured at 12000 $\mu e$. The radiograph of Specimen B-11 at 15000 $\mu e$ (Figure 23) show that several -45° cracks span the width. This, perhaps, explains the rise in the IST display from 14000 $\mu e$ to 14500 $\mu e$.

4. The magnitude of the modulus (or, for that matter, load) drop resulting when FPF occurs is 1/3 the magnitude of the drop found in AS4/3502 (see Figure 14). This signifies that the 90° plies in the AS4/3502 material carry a greater percentage of the total load.

5. Recognize that this specimen survived 15000 $\mu e$, a 9% increase over its static ultimate strain.

Experimental Results
Figure 21. Radiograph of Specimen B-7 at 12000 μs under IST loading.
Figure 22. Radiograph of Specimen B-10 at 13500 µs under IST loading.
Figure 23. Radiograph of Specimen B-11 at 15000 \( \mu \)s under IST loading.
The IST results of a “long-life” laminate (I-13) is shown in Figure 24. Highlights from this display include:

1. This test was started at 5000 με. The data at 5000 με was discarded since it did not register on the scale shown (the point was too large). This initial strain level is considerably lower than the strain level producing matrix cracking (see Point 2, below). The large drop in A modulus at 5000 and 5500 με may be due to the cumulative viscoelastic response. If the response is linear, then the response at 5000 με is equal to the sum of the responses which would have occurred had the ramp been incremented up to 5000 με. This, however, does not explain the non-monotonic response between the 5500-6500 με data points.

2. A very distinct peak is apparent from 9500 με to 10500 με. As in Specimen B-11, one could argue that FPF for I-13 would begin at 9500-10000 με. The IST display corresponding to Specimen I-12 (Figure 25) shows FPF around 9000 με. A radiograph of a specimen (I-8) at 9500 με (Figure 26) conveys the early stages of 90° failure. Perhaps for this specimen, FPF occurred at 9000 με.

3. The IST results of I-13 and I-12 (Figures 24 and 25, respectively) reveal a noticeable decrease in load drop at 11000 με. A “bump” (as opposed to a “peak”) appears between 11500-12000 με. Radiographs of specimens (I-4,I-7) at 11000 με (Figure 27) and 12000 με (Figure 28) shown no new damage patterns. Both Figures 27 and 28 would lead one to believe that the -45° plies do not fail prior to 14000-14500 με. A radiograph of a specimen (I-10) at 14500 με (Figure 29) shows the beginning of -45° cracking, yet, these cracks do not extend across the width. Specimen I-12 was strained at 100 με increments after reaching 14500 με. It survived 16400 με at a load of 9.71 kips, an increase of 10% over its static ultimate strain and 8% over its static ultimate load. Even at 16400 με, no significant -45° cracking takes place (see Figure 30).
Figure 24. IST output from Specimen I-13.
Figure 25. IST output from Specimen I-12.
Figure 26. Radiograph of Specimen I-8 at 9500 μs under IST loading.
Figure 27. Radiograph of Specimen I-4 at 11000 $\mu$s under IST loading.
Figure 28. Radiograph of Specimen I-7 at 12000 μe under IST loading.
Figure 29. Radiograph of Specimen I-10 at 14500 μs under IST loading.
Figure 30. Radiograph of Specimen I-12 at 16400 μs under IST loading.
A direct comparison of "long-life" responses between a B and I laminate (B-11 and I-13) is shown in Figure 31. The points worth noting include:

1. The magnitude of the modulus (load) drop, is, in general, greater in the I laminates compared to the B laminates. This is likely due to the presence of the interlayer. At high strains, the low-modulus interlayer should relax significantly. Though it carries only a minute portion of the total load, relaxing its share may account for difference between the magnitude of the I and B response.

2. Despite the offset present in the ordinate, there is surprisingly little offset in the abscissa. In this respect, the curves are strikingly similar.

3. Deviations after 14000 µε may indicate -45° cracking in the B laminate prior to its manifestation in the I laminate. X-ray results lend credence to this postulate.

The data from three separate tests on each laminate type were averaged to form a "master curve". Each point shown represents the average of at least two and, at most, three data points. The "master curves" for the B and I are superimposed for comparative purposes in Figure 32. Relevant observations include:

1. Despite the deviation from 9000 to 10500 µε, the two curves, for the most part, show little discrepancy along both the ordinate and abscissa.

2. Unlike the I curve, the B curve shows no distinct 90° "peak". Recall the "double peak" nature of the B laminates. Any offset along the abscissa in the three B tests might find "valleys" being averaged with "peaks", tending to wash-out both extrema. In this respect, the B "master curve" poorly represents the B data.
Figure 31. IST output from Specimen I-13 vs. Specimen B-11.
Figure 32. IST "master curves" of B and I response.
3. The formidable “peak” in the I curve reflects the small variability in the I response. The curve distinctly shows that 90° failure begins at 9000 με and progresses until 10500 με. Clearly, the 90° crack density is a function of the strain between those two limits.

4. No indication is given in either curve that -45° cracking has occurred prior to 14000 με.

A summary of the IST tests and test variables (defined on pages 33 and 35) is found in Table 8.

It is interesting to note that at least two specimens of each laminate type survived the average static failure load when subjected to IST loading. The limited data points prevent the formulation of any conclusions — be it that IST loading alters ultimate laminate response, or, that these points represent the upper-end of the strain variability.

One may draw some conclusions concerning the relative durability of the two laminate types. Durability is quantified during the IST by the area under the A modulus-applied strain “curve”. Excessive drop in modulus at each strain level reveals a lack of durability; the structure does not resist structural degradation well.

Looking at Figure 32, the claim is made that the disparity in the responses from 9000 με to 10500 με should be doubted. As stated earlier, the B response poorly represents the actual data since the averaging performed to create the “master curves” tends to “wash out” the variability of the response. Clearly, the responses outside of this strain interval are quite similar. Therefore, the respective area under each curve outside of this strain interval is similar. Since Figure 32 is not a reliable gauge for the average B response in the 9000-10500 με interval, a generalization is made from Figure 31. In nearly every instance, the amount of load drop is greater in the I laminate than in the B laminate at strains within the FPF regime. The difference in the response is attributed to the presence of the interlayer. Therefore, the total area under the I response (at a given strain above FPF) is greater than the total area under the B response. Since “durability” is narrowly defined as a structure’s ability to resist stiffness loss, on this basis, the presence of the interlayer tends to slightly lower the laminate’s durability.
Table 8. Summary of AS4/C985 IST Data.

<table>
<thead>
<tr>
<th>Spec.</th>
<th>B-10</th>
<th>B-11</th>
<th>B-12</th>
<th>I-11</th>
<th>I-12</th>
<th>I-13</th>
</tr>
</thead>
<tbody>
<tr>
<td>$e_0$ (µε)</td>
<td>6500</td>
<td>3000</td>
<td>3000</td>
<td>5000</td>
<td>4000</td>
<td>5000</td>
</tr>
<tr>
<td>$\Delta e$ (µε)</td>
<td>500</td>
<td>500</td>
<td>500</td>
<td>500</td>
<td>500</td>
<td>500</td>
</tr>
<tr>
<td>hold time</td>
<td>30</td>
<td>30</td>
<td>30</td>
<td>30</td>
<td>30</td>
<td>30</td>
</tr>
<tr>
<td>$e_f$ (µε)</td>
<td>13500</td>
<td>15000</td>
<td>14500 (fail.)</td>
<td>10000</td>
<td>16400</td>
<td>15200 (fail.)</td>
</tr>
<tr>
<td>$L_f$ (lb.)</td>
<td>8087</td>
<td>8753</td>
<td>9053</td>
<td>6094</td>
<td>9653</td>
<td>9258</td>
</tr>
<tr>
<td>$\Delta e'$ (µε)</td>
<td>-</td>
<td>-</td>
<td>-</td>
<td>-</td>
<td>100 @ 1.45%</td>
<td>100 @ 1.5%</td>
</tr>
<tr>
<td>purpose</td>
<td>AE</td>
<td>long-life</td>
<td>ultimate</td>
<td>AE</td>
<td>long-life</td>
<td>ultimate</td>
</tr>
</tbody>
</table>

$L_f$ is the highest load achieved
The data collected would seem to indicate that beyond FPF the I laminates "contain" the 90° ply damage more effectively than the B laminates. Matrix cracking in the -45° plies is seen at a lower strain level (around 14000 με) in the B than in the I laminates. "Damage tolerance" was earlier defined as the material's ability to resist failure over time once damage is present. Measuring the ultimate strain achieved in the IST, one may comparatively quantify the laminate's damage tolerance. The ultimate strains achieved by the I and B laminates reflect the same disparity in ultimate strains due to monotonic loading. The I laminates have a noticeably higher ultimate strain. Therefore, the interlayered laminates have fulfilled their original design purpose — they are more damage tolerant.

Experimental Results
Damage Analysis

The quantitative results of the preceding section indicate decisively that the interlayer plays an important role in the laminate's response to its loading condition. The presence of the interlayers enhanced the monotonic tensile response of the laminate, increasing the ultimate load by 8% for the AS4/C985 and 13% for the AS4/C1808 over their respective baseline laminates. Ultimate strain increases of 8% and 5% for the same materials were noted. The influence of the interlayer was quite obvious upon examining the notched fatigue response. The residual strength results indicate a decrease in tensile strength in the I laminates at an earlier percentage of total life compared to B laminates cycled at higher and lower load amplitudes. The IST highlighted several differences in the two laminate responses. The I laminates, again, revealed a larger ultimate load and strain than the B laminates. FPF in the two laminates was not decidedly different. Successive off-axis ply failures occurred at a more advanced strain level in the I laminates.

The mechanical tests have yielded the above results impassively. It is necessary to look "inside" the material with various nondestructive and destructive testing methods in order to explain or give reason to the results. In this manner, an examination is given as to "why" the above results were obtained for each loading condition. As before, each loading condition will be examined separately. It will be necessary, however, to "borrow" the results of the
other loading conditions in an attempt to codify a complete and thorough answer to the question, "why?"

**Monotonic Loading**

The language of the critical element scheme [49] is willfully incorporated into this discussion due, in part, to the clarity it introduces. "Critical" elements are defined as those elements whose failure translates into laminate failure. "Subcritical" elements are those whose failure does not directly cause laminate failure. Their decay promotes degradation of the critical elements, which, in turn, define the laminate's ultimate strength.

The discussion may be focused on the case at hand. For a quasi-isotropic laminate under monotonic tension, the "critical" elements are the 0° plies; failure of these plies is synonymous with laminate failure. The off-axis plies (90°, 45°, and -45°) are the "subcritical" elements; failure of these plies does not produce laminate failure. The strength of the critical elements may be accurately expressed using statistical variables. Therefore, given the stress state in these elements, one could formulate an expression for the probability of element failure at that stress level.

The introduction of interlayers into a quasi-isotropic laminate does not alter this notion. The 0° plies in each laminate type are comprised of the same material. The curing cycles of the I and B laminates are identical. Each material was exposed to identical moisture environments. There is no obvious reason why the 0° strength distribution should be different for the two laminates. It becomes apparent that the local stress state in the critical elements of the two laminates must differ in magnitude in accordance to the disparity found in ultimate "strengths" (where "strengths" should be the "adjusted strength", as defined earlier).

The magnitude of the respective local stress state is dictated by the mechanical integrity of the subcritical elements. The predominate damage mode of the subcritical elements in a
quasi-isotropic laminate loaded in monotonic tension is matrix cracking. These matrix cracks, characterized as through-the-thickness cracks running parallel to the fiber direction, usually initiate at the edge. Often cracks are found emanating from the edge without spanning the entire width of the specimen (see Figure 23). This is especially true for matrix cracks in the -45° and 45° plies. Advanced levels of load (strain) are required to "run" the crack across the entire width. Though other damage modes may exist (e.g., edge and interior delamination), little evidence of this is given from X-ray radiographs of failed specimens (see Figures 33 and 34).

It is assumed, then, that subcritical matrix cracking is the damage mode causing a significant change in the local stress state of the critical elements. For instance, when the 90° plies fail in a [0/45/90/-45]s laminate, the load once carried by those plies is redistributed to its neighboring plies (the 45° and -45° plies). Though the load transference is a local phenomenon, the summation of the load (strain) peaks over the entire length of these plies corresponds to a net increase in laminate strain. This increase is felt, ever so slightly, by the 0° plies. The same effect occurs when the -45° plies fail. The modeling of this load transfer process on the global behavior using CLT is known as the ply-discount scheme.

The failure of the 45° plies — the plies directly adjacent to the 0° plies — greatly affects the local stress state in the 0° plies. The 45° ply failure initiates zones of stress (strain) concentrations in the 0° plies due to load transfer — a direct consequence of the material inhomogeneity [78] — and due to the crack geometry itself. In addition, the matrix crack-tips have an associated "stress singularity" acting on the 0° plies as a function of the distance from the tip, r. This effect is independent of the material (provided that a true bond does exist between the lamina) and is a function solely of the crack-tip geometry.

Intimately connected to the notion of the crack-tip singularity is the concept of a microscopically small "edge notch". Given a pure bond between the lamina, one may assume the crack-tip to have penetrated an infinitesimal distance into the 0° ply since "infinite" peel stresses exist at its tip. If a local delamination occurs at the crack tip/0° ply interface, the
Figure 33. Radiograph of IST specimen B-12; failed at 14500 $\mu$s.
Figure 34. Radiograph of IST specimen I-13; failed at 15200 μs.
stress singularity is dissipated, crack-tip penetration does not occur, yet, the load transfer intensifies when the materials converge.

The structural decay of the subcritical elements may contribute additional load (that load above the simple mechanical load on the 0° plies) to the critical elements in these three ways:

1. the net increase in load due to the failure of the 90° and -45° plies.

2. the highly local concentrations of load due to matrix crack-tip stress singularities.

3. the local concentrations of load due to the geometry of the 45° cracks.

4. the local concentrations of load due to load transfer between the failed 45° plies and the 0° plies.

Each contribution will be assessed individually, hoping to distinguish the role of the interlayer. Load transferred to the 0° plies from 90° and -45° ply failures should not differ greatly between each laminate type. The amount of load transferred is not negligible. There is reason to believe, however, that it is approximately equal in each laminate. Performing a simple laminate analysis incorporating a ply-discount scheme (with laminated properties provided by American Cyanamid [74]) reveals that the 0° plies of the B laminate carry approximately 1% more load than the 0° plies of the I laminate after load transfer takes place.

A specimen from each laminate type (B-13 and I-14) was rapidly loaded to a near-failure strain (13500 and 14500 με, respectively), held at this strain for less than 5 seconds, and unloaded. The edge of each specimen was examined via photomicroscopy and edge replication. Each laminate exhibited comparable 90° crack density. This density was 2.047 cracks/mm in the B laminate and 1.850 cracks/mm in the I laminate. The -45° crack density showed a stunning discrepancy. The B laminate revealed 0.55 cracks/mm while the I laminate registered 0.079 cracks/mm. Though it is obvious that the -45° plies in the B laminate are shedding much more load, it is not established to what extent the 0° plies “see” this load.
From these arguments, the extent and degree of 90° and -45° matrix cracking in each laminate type will not be considered as a predominant influence on the local stress state of the 0° plies. Therefore, the 45° plies assume the role of “critical” subcritical elements. Surely, the 90° and -45° plies may be seen as subcritical elements to the 45° plies, inasmuch as their damage state may dictate the stress state in the 45° plies.

The X-ray radiographs of Specimens B-13 and I-14 (Figures 35 and 36) show that at advanced levels of strain, very little 45° matrix cracking is present. At such laminate strain levels, the addition of stress (strain) concentrations — be they from geometrical or load transfer sources — would be extremely threatening to the life of the 0° plies. Therefore, each 45° matrix crack may be deemed “critical” to the strength of the 0° ply. Yet, which stress concentration source poses the greater threat?

Singular stress fields (cf. [30,31]) at a crack terminus are characterized by infinitely large stresses at the crack tip that decay rapidly to the global field. Reifsnider [78] contends that — during quasi-static loading — local, singular stress concentrations play an insignificant role in the process towards final failure. He writes:

Based on some fifteen years experience in the composite community with the prediction of engineering strength of composite laminates, one can say for quasi-static loading that the region of influence of matrix cracking can be considered to be a region with a characteristic dimension that is of the order of magnitude of one ply thickness from the tip of a matrix crack; there is presently no evidence that the local stress singularity near the tip of the crack controls attendant damage development.

Stress concentrations due to the geometry of the crack may greatly influence the local stress state of the critical elements. Jamison in [79] reveals that the fiber fracture pattern resulting from off-axis matrix cracks resembles a “fan”; i.e., the highest density of fiber fracture occurs at or near the crack-tip, while, just a few fiber diameters away, little fiber fracture is found. No attempt was made in this study to quantify these influences.

Local stress concentrations due to load transfer at the 45° crack interfaces are presumed to significantly influence the critical elements during quasi-static loading. Differences in interlayered and baseline response shall be cast in this light.

Various schemes are often employed in an attempt to quantify the local influence of cracked plies on the adjacent load-bearing plies. One-dimensional shear-lag models [2,56],

**Damage Analysis**
Figure 35. Radiograph of monotonically loaded B-13 at 13500 μe.
Figure 36. Radiograph of monotonically loaded I-14 at 14500 μs.
finite-difference [35] and finite element [34,39] methods, and elasticity solutions [30,31], all varying in degree and complexity, represent the gamut of approaches towards this assessment. The 1-D shear-lag models lend considerable insight to the problem by virtue of their simplicity. The cracked ply (see Figure 37) transfers its load through a shear transfer region. This region has an associated shear stiffness, $G$, and a width, $b$, over which the transfer is confined. The width, $b$, is an arbitrary length, often chosen to “fit” the data. The ratio, $G/b$, effectively characterizes the shear transfer region; high ratios indicate an “Immediate” transference, i.e., load is returned back to the damaged ply a minute distance from the crack, while low $G/b$ ratios indicate that load lags back into the damaged ply after a considerable distance.

In baseline laminates, $G$ is often approximated by $G_{22}$ (since most shear transfer is assumed to take place within the resin-rich bondline), while $b$ is some fraction of the ply thickness, say 1/10. An interlayered laminate readily accommodates this analysis. The $G$ is simply the isotropic shear stiffness of the interlayer (provided the crack tip does not run completely through the layer) while the dimension, $b$, corresponds to the interlayer thickness (or, uncracked interlayer thickness).

Little persuasion should be necessary to convince the reader that $G/b$ of the interlayer is substantially less than $G/b$ of the baseline laminate. For the sake of argument, assume that the shear stiffness of the baseline resin-rich zone was equal to the $G$ of the isotropic interlayer. For the ratios to be equal, $b$ of the baseline must be equal to $b$ of the interlayer, a region that is, arguably, well defined. The interlayer thickness, 0.5 mil, translates into a B laminate shear-lag zone that is several fiber diameters thick. Therefore, the assumption that $G \leq G_{22}$ begins to appear erroneous since $G$ begins to physically approach $G_{22}$. Clearly, the $G$ in the B laminate should be larger than in the I laminate, while $b$ of the former is smaller than the latter. In other words, $G/b$ of the B laminate is greater than $G/b$ in the I laminate. It is very important to note that the shear transfer takes place, in part, in the adjacent lamina in all B laminates. In the I laminates, nearly all of shear transfer is presumably confined to the interlayer. This distinct, compliant material will accommodate strain concentrations by, likely,
Figure 37. Shear-lag analysis of a cracked ply.
enveloping the region in a large plastic zone. The influence this crack has on the adjacent, undamaged ply should be significantly less than in the B laminate scenario.

The interlayer, therefore, performs two crucial roles in the shear transfer region:

1. Nearly all shear transfer occurs in a distinct and compliant region. By possessing a comparatively low $G/b$, the interlayer causes the load to be gradually re-introduced into the damaged ply, and, subsequently, reduces the maximum peaks of strain and stress concentration in the adjacent undamaged ply.

2. The crack-tip is, usually, confined within the interlayer (see Figure 38). This allows for strain intensification within a ductile, non-critical region. Additional energy must be added to the system to move the crack to the undamaged ply interface; crack-tips in B laminates end either at or in the adjacent ply (see Figure 39).

These two assets have a cumulative effect as plies are progressively damaged. Returning to the results of B-13 and I-14 (Figures 35 and 36), can see that the 90° crack spacing is somewhat (though not much) larger in the I laminate than in the B. This difference should reflect the respective ratios of $G/b$. Yet, the -45° crack spacing is drastically different. This observation explicates the role the interlayer plays in isolating the undamaged ply from the damaged one. Figure 39 shows a 90° crack in a B laminate running along a path of high shear strain (the bond line) until it finds a "weakened area" where it may enter. The 90° cracks are "absorbed" by the interlayer (Figure 38) and -45° cracking is not promoted from the effects of these crack-tips.

The interlayer will tend to isolate the 45° cracks from the debilitating influence of the 90° cracks. Only a few (3 or 4) through-the-thickness matrix cracks were found in the 45° plies of Specimen B-13. It is important to remember that at the advanced level of strain (13500 με) each 45° crack is potentially threatening to the 0° plies. Fewer 45° cracks (one or none) were found in Specimen I-14 at 14500 με; a 7.5% higher strain than the strain reached by B-13. This clearly reinforces the concepts presented so far. The "critical" subcritical elements (the 45° plies) are more damaged in the B specimen at equal levels of strain. Compounding this trend
Figure 38. Photomicrograph of a 90° crack-tip confined in an interlayer.
Figure 39. Photomicrograph of a 90° crack-tip entering the adjacent ply in a baseline laminate.
is the fact that an interlayer still exists between the 45° plies and the 0° plies in the I specimen. In the B specimen, the 0° plies will ensure "immediate" shear transfer from any 45° matrix cracks due to their inherent stiffness; it is accomplished, however, at their expense. Very high strain and stress concentrations will accompany each 45° crack-tip. Fiber fracture caused by these concentrations will tend to be highly localized (cf. [36]) likely instigating a "domino effect", not unlike that proposed by Batdorf [80]. To have the same effect in the I laminate, cracks in the 45° plies would have to run through the interlayer. If this does not occur, the strain and stress concentrations in the 0° plies will be of comparatively less magnitude, allowing the 0° plies to attain a greater global strain prior to local, flaw-induced failures. As a consequence, interlayered laminates have a greater strain (and load) to failure than baseline laminates. Experimental results support this conclusion.

**Fatigue Loading**

Before addressing "why" the interlayer altered the fatigue response of notched, quasi-isotropic, AS4/C1808 laminates, it is necessary to briefly review the comparative test procedure and the relevant quantitative results.

The B laminates were cycled to two different, maximum load amplitudes; "high" (8000 lb.) and "low" (6200 lb.). The maximum load amplitude of the I laminates was 7300 lb. This "middle" load level nearly bisects the two B laminate load extremes. Questions may be raised concerning the comparability of the results obtained for the I laminate to the B results. Obviously, the "high" and "low" B responses are load dependent (see Figure 13 and Table 6). If a B laminate was tested at the "middle" load level (i.e., a "transition" load level), the response should be bracketed by the two extremes. The question to ask, then, is: "does the interlayer cause the 'middle' load response to deviate from the middle?"
The normalized stiffness vs. normalized life curves for the three load levels (Figure 13) gives every indication that the interlayer does not affect fatigue response; the “middle” load response is neatly bracketed by the two load extremes. The residual tensile strength data, however, does indicate a deviation from this “middle load” response. At “early” life, all the laminates revealed an increase in tensile residual strength over their monotonic values; this would correspond to the relaxation of the global stress concentration (the hole) due to damage. At “middle” life the situation changes, however. The B laminates at both load levels still show an increase in their tensile residual strength ratio. The tensile residual strength ratio of the I laminate, however, falls below 1.0. To understand why this occurs is to understand the effect interlayering has on the notched fatigue response.

X-ray radiographs of “early” life at each applied load level are shown in Figures 40-42. The intensity of off-axis matrix cracking is quite comparable in each case. The I laminate differs from the other two in the degree of 0° longitudinal splitting and by the amount delamination bordered by these splits. In the B laminates, at least 3 splits border each hole. Careful study will show that the delaminations present around the hole are “contained” in the lateral direction by these splits. In the I laminate, at nearly 19000 cycles (the same order of magnitude as the B laminate at low load), only one 0° split can be detected on each side of the hole. Again, the delamination stays within this confine. Reduced delamination is to be expected from I laminates; they were designed with this purpose in mind. It is apparent, however, that these slight differences do not greatly affect the residual tensile strength at this percentage of total life.

Figures 43-45 show radiographs of “middle” life laminates at each applied load level. These radiographs show three distinct responses. The B laminate under high load (Figure 43) shows a considerable concentration of damage near the hole. Matrix cracking is sparse outside of this intense band of damage. The acute localization of damage around the hole has reduced the global stress concentration caused by the hole [46]. Therefore, the laminate residual strength increases. The edge radiograph from Fig. 45 shows the inner 0° plies as distinct white bands. Upon careful examination one can see dark lines — indicating delami-
Figure 40. Radiograph of a B specimen at early life, high load; $n = 904$ cycles.
Figure 41. Radiograph of an I specimen at early life, middle load; $n = 18882$ cycles.
Figure 42. Radiograph of a B specimen at early life, low load; $n = 25000$ cycles.
nation—bordering each side of the 0° plies. Large strain concentrations will arise at the hole due to the geometry. Delamination near the hole, isolating the 0° plies, will "free" the 0° plies from a triaxial stress state in the debond area, yet, will cause a local stress concentration where the material rejoins. By reducing the constraint at the hole, the 0° plies remain intact.

The damage state in the I laminate (Figure 44) is markedly different. Matrix cracking in the 0° plies extends on each side of the hole over a distance greater than the hole radius. Off-axis matrix cracking spans the width of the specimen, confined within the two criss-crossing lines ~45° to the horizontal. It is interesting to note the symmetrical, "ear-shape" pattern of dense matrix damage extending from the hole. The width of these "ears" is approximately the diameter of the hole. It is striking how similar these patterns resemble the images obtained from thermoelastic measurements of isotropic and quasi-isotropic center-notched materials [81].

The most important damage features in this specimen occur at or near the hole boundary. Delamination is present, yet, it is confined to a narrow region. The dark, triangular regions of intense damage give every indication that large-scale ply fractures have occurred within this region. The edge radiograph repudiates this supposition. Three to four distinct ply failures are visible. The distinct "bands" of 0° plies seen at the top of the specimen are not isolated as being "damage free" in the hole boundary. This extensive ply damage, as seen in the edge radiograph, tends to extend the hole damage in a direction transverse from the load axis. Reifsnider, et al., state in [46] that, "tensile and compressive strength reduction appears to be controlled by damage growth along the axis transverse to the loading direction, through the center of the hole." The damage present in this I specimen dictates a reduction in the residual tensile strength.

Before postulating the causes of this intense, debilitating damage, the damage state of the B laminate under low at "middle" life (Figure 45) will be discussed for comparative purposes. Damage in the form of off-axis matrix cracking and delamination is much more extensive than the other two laminates at this stage of life. Though the magnitude of the damage appears ominous, no particular zone of intensity seems life-threatening. Triangular regions of local
Figure 43. Front and edge radiograph of a B specimen at middle life, high load; \( n = 9240 \) cycles.
Figure 44. Front and edge radiograph of an I specimen at middle life, middle load; $n = 110000$ cycles.
Figure 45. Front and edge radiograph of a B specimen at middle life, low load; $n = 1052000$ cycles.
damage flank each side of the hole, however, in comparison to Figures 43 and 44, this region does not appear as damaged. The edge radiograph reveals damage at each surface. This surface damage alters the strain measurement across the hole causing a sharp “dip” in the stiffness curve (see Figure 12). It is surprising that the tensile strength did not degrade in this specimen since the net section area in which the load is carried has been reduced. Recognize the white “bands” of 0° plies within this laminate. The core 0° plies remain intact, as was the case in Figure 43, though apparently not so in Figure 44.

The I laminate, cycled at a “middle” load level, shares damage patterns characteristic of each of the load extremes. Like the B laminate under high load, its damage is localized near the hole boundary. Like the B laminate under low load, its off-axis matrix damage extends to the width of the specimen. One obvious distinction displaces its performance as merely the “average” of the two extremes; its tensile strength degrades at an earlier percentage of total life. The reason for this has to do with the intensification of damage at the hole, with the tendency to extend this damage towards the width of the specimen. Damage development parallel to the load axis serves to diminish the geometrical stress concentration of hole [46]. Comparing Figure 43, a high-load laminate, to Figure 44, one can see that in the former case the damage extends above and below the hole more so than the latter. This is the damage that tends to be “beneficial” to notched tensile residual strength.

The interlayer's ability to resist delamination is likely the cause for the laminate’s degradation in tensile strength. The distinct “bands” of 0° plies found in Figures 43 and 45 are seemingly isolated by the delaminations that appear on each side of the plies. As mentioned earlier, these delaminations at the hole boundary relieve the 0° plies — already under severe geometrical stress concentrations — from the onus of a triaxial stress state. In the I laminate, the lack of large-scale delamination likely exposes the 0° plies to a triaxial stress state, causing local ply failures. These local failures, may, in turn, instigate more matrix cracking and delamination [82]. Yet, perhaps the damage containment ability of the interlayer does not allow this local stress relief to occur.
It should be stated that this damage scenario is, at best, theoretical supposition. Razvan [83] has produced informative results by utilizing the deeply technique in conjunction with scanning electron microscopy to determine the causative sequence of delamination and ply fracture in the notched, baseline, AS4/C1808 laminates. Such work, if applied to the I laminates, would end much supposition and yield an insightful look at the boundary-region damage processes.

The late-life radiographs of the "high", "middle", and "low" load specimens are found in Figures 46-48. Again, the I laminate shares characteristics of both extremes. The extent of its damage above and below the hole is reminiscent of the low-load specimen. The region of intense damage at the hole boundary is not unlike the high-load specimen. Comparing the edge radiographs highlights the similarity in the damage of the "late" life specimens. Each reveals gross failure beginning at one surface. The unseen difference is in the failure modes. The B laminates fail in a compression/instability mode despite their respective applied load level. The I laminates fail in a tensile mode (at this particular load level). Gross delamination tends to be the culprit causing compression failures. In a delamination-resistant laminate, it is not surprising that tensile strength degrades more rapidly then compression strength.

The tensile performance during fatigue cycling (R=-1) of the notched, damage tolerant (interlayered) laminate was found to be inferior to the baseline laminates at the particular load level tested. This finding concurs with the results of [43] and the tendencies mentioned in [84]. One should be reminded that the compressive performance was not monitored. Though generalizations may be drawn, research should follow-up on this notion. Similarly, one must be reminded that the I laminate was tested at one load level. Drawing analogies from the work of Simonds, et al. [43], one might imagine a drastic load-level transition to occur in this material. In other words, testing the I laminates at a low-load level may have resulted in superior performance versus the baseline laminate in both total life and strength retention throughout life. This, however, is left to future research.
Figure 46. Front and edge radiograph of a B specimen at late life, high load; $n = 8800$ cycles.
Figure 47. Front and edge radiograph of an I specimen at late life, middle load; \( n = 383310 \) cycles.
Figure 48. Front and edge radiograph of a B specimen at late life, low load; \( n = 2215030 \) cycles.

Damage Analysis
Incremental Strain

The IST could be thought of as a monotonic test in which the effect of time on the laminate response is explored. At one end of the spectrum, as the time interval in which strain is held constant is decreased to zero (see Figure 2) the test becomes purely monotonic. To accentuate the effect of time on the material's response, however, one might ramp the strain at a very low increment (1 μs, for instance) and hold this increment until load equilibrium is achieved. If the material's response is a function of applied strain (stress) and not a function of time, then $\Delta \bar{A}$ should be zero at all $\epsilon$; the damage occurs instantaneously upon ramping. IST results indicate that $\Delta \bar{A}$ is seldom non-zero. Therefore, the laminate response is, indeed, time-dependent.

The "interrupted-ramp strain input" of the IST superimposes relaxation loading upon monotonic loading. It is necessary to distinguish the response due to the relaxation loading since the monotonic response has been studied. These questions should be addressed:

1. How and why is the IST response different from the monotonic response?

2. Why is the interlayered IST response different than the baseline response?

A discussion will be devoted to each point.

Prior to attaining FPF strains in the IST, the baseline response should not be dramatically different from its monotonic response. Perhaps some viscoelastic deformation occurs in the 90° plies, yet, at such a low applied strain, it should not be very significant. The distinction between the loading conditions materializes after FPF occurs. Clearly, stretching a laminate to a constant level of ply-failure strain encourages time-dependent cracking. Though the decreasing laminate compliance will reduce the applied load, the next increment of strain will initiate more failure, re-initiating the failure process. In this manner, the IST promotes a saturated damage condition (or, characteristic damage state) for any particular low-strength ply.
In monotonic loading, the failure is likely strain-dependent more so than time-dependent. Therefore, *damage accumulation* distinguishes the IST response from the monotonic response at strain levels above FPF in the material systems studied in this report.

The rationality developed for unnotched, tensile-loaded coupons indicated that ultimate load is a function of matrix damage, i.e., if matrix damage can be retarded, strength is expected to increase. If the IST promotes damage accumulation then these laminates *should* be weaker. Compare Table 8 to Table 2. One specimen each of the I and B laminates showed a tremendous increase in ultimate behavior during the IST. Apparently, IST loading *promotes* ultimate behavior. This contradiction must be resolved.

Figures 33 and 34 shed a great deal of light on this dilemma. Both the I and B specimens show a preponderance of 90° ply damage away from the failed surface. In addition, the B specimen reveals some -45° cracking. In each case, however, both the -45° and 45° plies were *far* from saturation damage states. Damage accumulation is debilitating to a laminate because:

1. the undamaged plies take on a greater portion of the load, and,

2. the presence of the damage and its associated stress concentrations jeopardizes the undamaged laminate. Statistical considerations illuminate this point. The more matrix cracks, delaminations, etc., the more local stress concentrations. A greater number of stress concentrations increases the probability of laminate failure.

Compare Figures 22 and 29 to 35 and 36. One can see that 90° crack density in the I and B laminates loaded in the IST is greater than the specimens loaded monotonically. As stated earlier, 90° ply failure has a minimal effect on the critical element (0° ply) stress state. More -45° cracks are present in the monotonic B specimen than the IST B specimen, yet, these crack states are far from saturation. Since the damage state of the respective failed surfaces (Figures 33 and 34) is not unlike those in Figures 22 and 29, the time-dependent degradation of the 45° plies is not a significant damage process in either laminate. In view of the failure
damage states, IST loading need not be debilitating to ultimate tensile behavior. Yet, why would IST loading seemingly promote ultimate behavior?

The philosophy posited until now has been deductive; a sequence of events have been formulated from the empirical results. Without exception, the formulation would have benefitted greatly from fractography studies. Examination of deplled failed and near-failed specimens under a SEM would have yielded considerable information. Without this analysis, however, speculative thinking must commence.

The mechanisms of dynamic crack propagation may explain the disparity in ultimate performance between monotonic and IST loaded specimens. The effect of loading rate on homogeneous materials is readily conceptualized. At high loading rates the fracture process is often dominated by one major “macrocrack”. At very low loading rates, defects appear and coalesce into cracks. The cracks grow, combine, and form into macrocracks. The “accumulation” of this damage will likely encounter an area of weakness. One of the many cracks will run until failure. The more damage present, the more likely that failure will occur. For a nonhomogeneous material the scenario is not as intuitive. At high loading rates, failure is, again, assumed to be initiated by one major flaw. In a layered composite, each layer inherently possesses a resistance to damage (quantified, conventionally, as $G_c$). Suppose a crack travels easily through 45° ply and is approaching the 45°/0° interface. Assume, also, that much more energy is required to run the crack in the 0° ply than in the 45° ply. If the crack is stable — with a minimal release of energy accompanying its growth — it will likely halt at the interface. However, if the crack is dynamic — possessing a large “crack-driving force” — then it is likely to bridge the interface. It does not, necessarily, run completely through the 0° ply, yet, if it enters the ply it is potentially threatening to the laminate’s strength. The claim now made is that monotonic loading results in “dynamic crack” failure caused by a major flaw. IST loading, however, likely produces more stable cracking. A major flaw created from the accumulated damage eventually contains the energy sufficient to enter and instigate 0° failure. The applied strain (and load) when this occurs may be distinctly higher than the applied monotonic strain producing laminate failure. The two specimens that failed during IST
loading — at applied strains above their respective average, monotonic ultimate strain —
 maintained a load above their failure load for a definite period of time. Specimen B-12 failed
 1 minute after 14500 μe was applied; l-13 failed 20 minutes after reaching 15200 μe.

In a series of papers [2,85-87] Suvorova, et al., have published data purporting a non-
monotonic relation between “orthogonally reinforced” composite laminate strength and loading
rate. At very slow loading rates, laminate strength is a linearly increasing function of rate.
Failure in this regime, they report, is due to damage accumulation. A loading rate is reached,
however, where, with increasing rate the relationship becomes linearly decreasing, i.e.,
strength goes down as the rate increases. This regime signifies the change into a failure
mode penned “macrocrack propagation”. Another inflection point is achieved as the loading
rate further increases. A linearly increasing relationship between strength and rate is resumed.
A new failure mode is associated with this change in slope; it is deemed a “cleavage
failure”. Tests performed by Daniel, et al. [88], on unidirectional specimens does not reveal
such an erratic strength-loading rate dependency. If the unusual results of Suvorova and his
co-workers are to be believed, considerable credence is lent to the complicated hypothesis
posed above.

The strain level producing FPF in each laminate type is atypically similar. The
interlayer’s ability to change the local, three-dimensional stress state should seemingly en-
sure that FPF occur at a higher strain level. Matrix cracking in the 90° plies, with few ex-
ceptions, originates from the free-edge. Often times machine damage contributes to “starter
notches,” though it is this researcher’s experience that polishing does little to relieve the sit-
tuation. Damage originates from the free-edge because of the intense 3-D stress fields pres-
ent. Kistner, et al. [89], report that the axial stresses may increase by 250% (depending on
the lay-up) due to the 3-D boundary layer effect. While the 2-D stress state in the 90° plies of
each laminate type prior to FPF is quite similar (since lamina of equal stiffnesses are sub-
jected to equal applied strains), clearly the out-of-plane stresses are not similar. Interlayers
were developed expressly to lower these stresses. On the basis of this argument it is sur-
prising that the onset of FPF does not occur at a distinctly higher strain level. Perhaps this
argument is more appropriate in view of -45° ply failure; Figures 22 and 29 show conclusively that -45° matrix cracking begins at a substantially lower strain level in the B laminates.

The hypothetical framework established in order to explain the enhanced IST ultimate behavior easily accommodates the distinction between I and B IST ultimate response. The compliant layer between the 45° and 0° plies serves as a "crack arrester," dissipating the crack-tip energy of a particularly threatening macrocrack. In this manner, it is not entirely surprising that the I laminate's enhanced ultimate response during IST loading reflects the same improvement that the I laminate possesses over the B laminate with respect to ultimate monotonic response.
Conclusions and Recommendations

The focus of this work addressed the following two questions:

1. Does the presence of the interlayer alter the laminate's response to its loading condition?
2. If so, why?

The answer to the first question simply involved the quantitative comparison of interlayered and baseline results. It was determined that the interlayers did, indeed, alter the laminate's mechanical response. Various techniques were used to identify the damage mechanisms present. From this data, arguments were constructed to explain why the interlayer had the effect it did. In this chapter, relevant findings and conclusions that contributed to the problem above are summarized. In accordance to the form already established, this chapter shall be apportioned into the three loading conditions investigated.
Monotonic Loading

Experimental Results

AS4/C985

1. Three monotonic tensile tests to failure were performed on baseline and interlayered, 16-ply, quasi-isotropic, AS4/C985 laminates. On average, the interlayered laminates revealed an 8% greater load and strain at failure.

2. The interlayer, 0.5 mil each in thickness, appeared between each ply and on one outside ply. It contributed, on average, 6% to the baseline thickness and 6% to the laminate’s mass.

3. The average failure strain of 14930 με recorded for the interlayered laminates seems remarkable when one considers that AS4 fiber is listed as having a failure strain of 1.5%.

4. Comparing the strengths obtained from the I and B laminates can be deceiving. The thin, low-modulus interlayers carried a minute fraction (~ 1%) of the total load, yet, contributed 6% to the cross-sectional area. If the strength is adjusted to reflect this discrepancy then the I laminates show a 7% increase over the baseline strength.

5. The I laminate stiffness was 5% less than the baseline stiffness. This decrease in stiffness may be approximated from CLT if lamina data and interlayer properties are known.
6. In general, the interlayered laminates had very little variability in their ultimate performance.

7. Little delamination was recorded in both laminates at or near failure.

8. Monotonic compressive results for the AS4/C985 laminates were not obtained.

AS4/C1808

1. Two monotonic tensile tests to failure were performed on interlayered, 32-ply, center-notched, quasi-isotropic, AS4/C1808 laminates. On average, the load at failure revealed a 13% increase over the baseline laminate (performed by another investigator), while the strain to failure increased by 5%.

2. The interlayer material present in the AS4/C985 laminates, again, appeared between each ply of the AS4/C1808 laminates. The interlayer contributed 9% to the laminate thickness.

3. An "adjusted" strength was not calculated for the I laminates due to insufficient or questionable data.

4. On average, the I laminate showed a 10% decrease in axial stiffness. Stiffness values were approximated from the X — Y plots of load vs. strain; strain was measured across the center-hole. Some doubt is cast on the quantitative precision of the stiffness values.

5. Two monotonic compressive tests to failure were performed on I laminates and compared to baseline results. The ultimate I data showed little disparity between tensile and compressive performance. The baseline results increased significantly in compression. The average baseline load at failure rose 16% over its tensile result, while the strain at failure rose 49%.

Conclusions and Recommendations
6. The question remains whether the baseline, center-notched, AS4/C1808 laminates' compressive data were anomalous, or, the presence of the interlayer debilitated the monotonic compressive performance.

**Damage Analysis**

1. Radiographs of failed and near-failed I and B laminates revealed that laminate failure occurs prior to saturation of damage in the -45° and 45° plies. In both laminate types little to no 45° ply damage was apparent prior to laminate failure.

2. Photomicrographs revealed that the interlayers are effective at terminating ply cracks. It was surmised that the compliant interlayers tended to reduce the geometrical and local stress concentrations produced by adjacent-ply matrix cracks.

3. The “buffer zone” created by the interlayers between the “critical” 0° plies and the “subcritical” [45/90/-45]_s sublamine exploits its shear transfer and crack blunting capability and effectively lowers the local, intense, stress concentrations caused by the subcritical damage. In this manner, ultimate performance is expected to improve.

4. Fractography studies of deplied specimens using a SEM would greatly contribute to the understanding of the I and B failure process. Fiber failure patterns in the 0° plies of each laminate type could be compared. Those results would help substantiate the claims posited.

5. Tests to determine the effect of temperature on the comparative monotonic performance of the I and B laminates have been attempted, yet, none of the results are reported here. The data from such tests has extreme relevance to the current thrust of producing reliable, high-temperature, structural components.

Conclusions and Recommendations
Fatigue Loading

Experimental Results

1. Interlayered, 32-ply, center-notched, quasi-isotropic, AS4/C1808 laminates were subjected to fully-reversed (R = -1), tension-compression fatigue cycling. Specimens were cycled at a maximum load amplitude of 7300 lb., bisecting the two maximum load amplitude extremes (6200 lb. and 8000 lb.) used in the baseline fatigue test matrix (performed by another investigator).

2. Fatigue tests on each laminate type were run to failure in an attempt to estimate their respective fatigue life. The life of the high-load B laminates was ~10,000 cycles while the low-load life was ~1 million cycles. The "middle load" I laminate's fatigue life was ~400,000 cycles.

3. Stiffness measurements were taken at various fractions of the laminate's total life. The stiffness response of the I laminate fell neatly between the two B laminate extremes.

4. Laminates fatigued to "early," "middle," and "late" life were tested for residual tensile strength. All laminates exhibited an increase in strength at early life due to the relaxation of the center-notch stress concentration. The I laminate's residual tensile strength dropped, unlike the two B laminates at "middle" life. At "late" life, the I laminate maintained its previous level of residual strength, as did the low-load B laminate, while the strength of the high-load B laminate fell.

5. The results of the residual strength testing indicated that the presence of the interlayer did effect the laminate's fatigue response.

Conclusions and Recommendations
6. No attempt was made to cycle the I laminates at any other load amplitude. Recent published results [43] indicate that the I laminates might exhibit better performance at low-load amplitudes, while a transition in performance may be expected during high-load cycling. This possibility warrants further study.

7. Compression residual strength was not monitored, so no conclusions may be drawn. Again, much insight to the problem at hand could be gleaned from the results of such testing.

**Damage Analysis**

1. Edge X-ray radiographs lent a great deal of insight into the mechanisms involved in reducing I laminate residual tensile strength at "middle" life. In the B laminates, delamination was seen on either side of the "core" 0° plies at the notch. These delaminations served to isolate the plies, i.e., they were noticeably free from damage. The "core" 0° plies in the I laminate, on the other hand, were, seemingly, not isolated from the rest of the laminate. 0° ply fractures were apparent and likely promoted the decay of laminate tensile strength.

2. The interlayer's ability to resist delamination was seen as the culprit in this very unique instance. Delamination around the global stress concentration (the center-notch) tends to reduce the intense 3-D stress state experienced by the local elements. If continuity is maintained by these elements, the brunt of the 3-D stress field is experienced, likely promoting fiber failure. This fiber failure, in turn, lessens the laminate tensile strength.

3. These results concur with the generalizations arrived at by Simonds, et al. [43].

*Conclusions and Recommendations*
4. SEM analysis of deplied specimens would be an ideal tool for this particular study. Concrete evidence concerning the fiber failure process could be provided by such a study.

**Incremental Strain Loading**

The IST Procedure

1. The IST serves as an easy, accelerated, long-term test providing the variability to allow the researcher to tailor the test to his/her specific needs.

2. The "interrupted-ramp strain input" combines the aspects of monotonic and relaxation loading. The test exploits the time-dependent failure processes that are present within a given laminate.

3. By monitoring the change in "apparent modulus" per each increment of applied strain, a "curve" may be output highlighting the strain intervals causing significant structural changes. "Peaks" in the curve have been successfully correlated to the presence of off-axis matrix cracking through the use of X-ray radiography and acoustic emission.

4. The test method was shown to produce an acceptable level of reproducibility.

5. "Durability" may be comparatively quantified during the IST by measuring the area under the A modulus - applied strain "curve" (when considered as a histogram).

6. "Damage tolerance" may be comparatively quantified by measuring the ultimate performance of a specimen during the IST.
7. By manipulating the variables in the computer-controlled test system, strain rate effects may be explored.

8. The test procedure readily accommodates high-temperature testing. Initial attempts at performing this task have met with some success. These results are not reported here.

9. The IST is presented to the engineering community as an effective and useful tool in the study of damage in composite laminates.

Experimental Results

1. Both laminate types yielded spurious IST data at low (< 5000 µε) applied strains. This may be attributed to system noise or other system deficiencies.

2. Prior to FPF each laminate type revealed a marginal amount of stiffness loss. It is inconclusive whether this loss may be attributed solely to viscoelastic creep. No separate attempt was made to quantify this phenomenon.

3. The strain level in which FPF occurred is not distinctly different between the two laminate types. FPF began between 9000-9500 µε for both the I and B laminates.

4. During the strain interval in which 90° matrix cracking proliferates, relaxation occurred in the I laminates to a greater extent than the B laminates. This may be due in part to the deformation arising in the interlayers as they assume the role of a "shear transfer region." Since the interlayers do not carry much load, this effect is not expected to be significant.
5. Matrix cracking in the -45° plies occurred in the B laminates at a lower strain level than in the I laminates (~13500 με compared to >14500 με).

6. The ultimate load and strain of the I laminates loaded in the IST was significantly higher than the ultimate B IST results. The increase in performance was quite similar to the increase found during monotonic loading. Ultimate performance in both laminate types was slightly enhanced during IST loading compared to monotonic loading.

7. From the definitions posited, the I laminate exhibited lower durability than the B laminate. However, the I laminate was deemed more damage tolerant; despite the damage present at high applied strains the laminate effectively "contained" that damage and maintained its integrity.

**Damage Analysis**

1. One would expect a greater preponderance of 90° ply cracking during IST loading than during monotonic loading. Holding the laminate under constant strain allows for considerable time-dependent cracking. Depending on the IST loading rate, 90° ply damage was allowed to fully saturate. Examination of X-ray radiographs of both laminate types showed that the 90° crack density was, indeed, greater in the IST-loaded laminates.

2. Little to no 45° ply cracking was recorded in IST-loaded laminates prior to failure. This concurred with the X-ray results obtained during monotonic loading.

3. Mechanisms of crack propagation may explain the disparity found between the IST and monotonic ultimate behavior. Hypothetically, "dynamic" macrocrack propagation could control monotonic tensile strength. This would explain the rise in I laminate monotonic strength since the interlayers would dissipate crack-tip energy. IST ultimate failure may
be initiated by "stable" crack growth; i.e., higher loads may be attained before life-threatening damage violates the 0° plies. This scenario could, again, account for the enhanced I laminate IST ultimate performance.

4. Obviously, SEM analysis of deplied specimens would greatly contribute to the validation (or rejection) of the elaborate, hypothetical processes proposed.

5. High-temperature IST studies have been under way, yet, no results are reported herein. The import of such an analysis is obvious. Long-term testing within a hot, moist environment would represent a "worst-case scenario." If interlayered laminates could prove their structural worthiness under those conditions, the design community would embrace the concept whole-heartedly.
References


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75. Tsai, S. W. and Hahn H. T., Introduction to Composite Materials, Technomic, 1980, pp. 73-74.

76. Bakis, C. E., private communication

References


Appendix A.

Two computer program listings are included herein. The first listing corresponds to the Incremental Strain Test. The program was written in BASIC. The second program performs the acoustic emission data manipulation from the IST data. This program was written in FORTRAN.

Incremental Strain Test

10 CLS
20 KEY OFF
40 COLOR 4,0
50 SOUND 800,36 : PRINT: PRINT " ** WARNING! DO NOT CONNECT MACHINE TO BOARD UNTIL NOTICE IS GIVEN! ** "
60 COLOR 7,0
70 ' PCLAB setup
80 ' ADC.VALUE=3 'entry point for routine
90 ' DAC.VALUE=24 'entry point for routine
100 DEF SEG=&H0 ' get the PCLAB segment
110 PCLSEG = PEEK ( &H4FE ) + 256*PEEK ( &H4FF )
120 DEF SEG=PCLSEG ' REM address the PCLAB segment
130 ' Program input
140 ' 170 '
180 PRINT
190 PRINT "Which program is to be run?":PRINT
200 PRINT "1. RELAXC"
210 PRINT "2. RELAXT"
220 PRINT "3. RELAXG"
230 PRINT "4. RAMPS":PRINT
240 INPUT "Enter the number of the program: ", PROG
250 PRINT
260 INPUT "Will AE testing be performed (no = 0, yes = 1)? ", AE
270 PRINT
280 INPUT "Enter the cross-sectional area of the specimen (sq.in.)? ", CROSS.AREA
290 PRINT
300 INPUT "Enter the load range (100, 50, 20, 10)? ", LOAD.RANGE
310 PRINT
320 INPUT "Enter the strain range (100, 50, 20, 10)? ", STRAIN.RANGE
330 PRINT
340 INPUT "Enter the starting strain level (in microstrain)? ", STRAIN.START
350 PRINT
360 INPUT "Enter the final strain level (in microstrain)? ", STRAIN.END
370 PRINT
380 INPUT "Enter the strain increment (in microstrain)? ", STRAIN.INC
390 PRINT
400 INPUT "Enter the time between data sampling (in sec.)? ", TIM.INC
410 PRINT
420 INPUT "Enter the time between ramping strain (in sec.)? ", RAMP.INC
430 IF INT(RAMP.INC) MOD INT(TIM.INC) <> 0 THEN PRINT "MUST BE A MULTIPLE OF SAMPLING TIME!":GOTO 420
440 PRINT
450 IF AE=1 THEN DIM AERMS(1000)
460 IF PROG=3 THEN Q= ((120/TIM.INC)-1) : GOTO 570
470 INPUT "Do you wish to store the data on a disk (Y/N)? ", D$
480 IF D$="N" OR D$="n" THEN GOTO 570
490 PRINT
500 INPUT "Input the filename (disk.filename.filetype)? ", E$
510 OPEN E$ FOR OUTPUT AS #1
520 IF AE=0 THEN GOTO 550
530 WRITE #1, "TIME STRAIN LOAD MODULUS DMOD AVG.RMS"
540 GOTO 550
550 WRITE #1, "TIME STRAIN LOAD MODULUS DMOD"
560 CLOSE #1
570 PRINT
580 ' Initializing data
590 ' Highest voltage in range.
600 ' Lowest voltage in range.
610 RANGE! = HIGH.V! - LOW.V! ' Total voltage range.
620 HIGH.V! = +10!
630 LOW.V! = -10!
640 ' Voltage of Least Significant Bit.
650 NOC! = 4096
660 ' Counts Per Volt.
670 LSB! = RANGE!/NOC!
LOADS.PER.VOLT! = LOAD.RANGE * 20
710 STRAIN.PER.VOLT! = STRAIN.RANGE * .0002
720
730 INT.VOLT! = (STRAIN.START*.000001)/STRAIN.PER.VOLT!
740 FIN.VOLT! = (STRAIN.END *.000001)/STRAIN.PER.VOLT!
750
760 INITAL.STRAIN! = (INT.VOLT! - LOW.VI) * CPVI
770 INITAL.STRAIN% = INITAL.STRAIN!
780 TERM.STRAIN% = INT((FIN.VOLT! - LOW.VI) * CPVI)
790 STRAIN.INC.DIGIT! = STRAIN.INC / 9.76
800 FIRST.STRAIN.DIFF! = INITAL.STRAIN! - 2048
810
820 DAC.SELECT% = 0
830 CHANEL.ZERO% = 0
840 CHANEL.ONE% = 1
850 CHANEL.TWO% = 2
860 GAIN.ONE% = 1
865 GAIN.TWO% = 2 : LSB.AE! = LSB!/GAIN.TWO%
870
880 STRAIN.SAMP% = 0 : LOADS.SAMP% = 0
890 STRAIN.RESET% =2048
900 I=1 : H=0 : INC=1 : Z=0 : F=0 : T(0) = -3 : L=2 : R=0
910 ITER% = (RAMP.INC / TIM.INC) + 1
920 IF PROG=2 THEN ITER.INC% = ITER%
930
940 CALL DAC.VALUE (DAC.SELECT%,STRAIN.RESET%)
950
960 COLOR 4,0
970 SOUND 800,36 : PRINT : PRINT ** FIRST INSURE THAT
980 SOUND 800,36 : PRINT : PRINT ** THE SPAN IS SET TO ZERO ***
990 SOUND 800,36 : PRINT : PRINT ** THEN YOU MAY CONNECT
1000 PRINT: PRINT** THE MACHINE TO THE BOARD ***
1010 PRINT: PRINT ** AFTERWARDS TURN THE SPAN TO
1020 IF AE=0 THEN GOTO 1080
1030 PRINT " TIME STRAIN LOAD MODULUS DMOD
1040 PRINT " TIME STRAIN LOAD MODULUS DMOD
1050 IF AE=0 THEN GOTO 1080
1060 PRINT " TIME STRAIN LOAD MODULUS DMOD
1070 GOTO 1090
1080 PRINT " TIME STRAIN LOAD MODULUS
1090
1100 End initializing
1110
1120 IF PROG=3 THEN ON ERROR GOTO 2860
1130
1140 TIMES$= "00.00.00"

Appendix A.
1150 \* 
1160 IF I=1 THEN GOTO 1270
1170 \* 
1180 \* Routine to ramp to increment of microstrain
1190 \* 
1200 IF STRAIN.DES! >= TERM.STRAIN% THEN GOTO 2400
1210 H = H + 1
1220 STRAIN.DES.VOLT! = ((STRAIN.INC "H" .000001)/STRAIN.PER.VOLT!) + INT.VOLT!
1230 STRAIN.DES! = ((STRAIN.DES.VOLT! - LOW.VI) * CPV!)
1240 Z=0 : L=2
1250 GOTO 1460
1260 \*
1270 IF CINT(FIRST.STRAIN.DIFF!) = 0 THEN FINAL.STRAIN! = 2048 + STRAIN.INC.DIGIT! ELSE FINAL.STRAIN! = INITIAL.STRAIN! : INITIAL.STRAIN% = 2048 : H = -1
1280 STRAIN.DES! = FINAL.STRAIN!
1290 FINAL.STRAIN% = FINAL.STRAIN!
1300 H = H + 1
1310 \* 
1320 FOR STRAIN.TO.APPLY% = INITIAL.STRAIN% TO FINAL.STRAIN% STEP INC
1330 CALL DAC.VALUE (DAC.SELECT%, STRAIN.TO.APPLY%)
1340 IF AE=0 THEN GOTO 1400
1350 CALL ADC.VALUE (CHANNEL.TWO%, GAIN.TWO%, AERMS.SAMP%)
1360 TOTAE = AERMS.SAMP% + TOTAE
1370 CT = CT + 1
1380 FOR K = 1 TO 10: NEXT K
1390 GOTO 1410
1400 FOR K = 1 TO 30: NEXT K
1410 NEXT STRAIN.TO.APPLY%
1420 INITIAL.STRAIN% = FINAL.STRAIN% : INC = 1
1430 \* 
1440 \* Routine for strain feedback and offset
1450 \* 
1460 FOR M = 1 TO 16
1470 CALL ADC.VALUE (CHANNEL.ZERO%, GAIN.ONE%, STRAIN.SAMP%)
1480 TOTSAM = STRAIN.SAMP% + TOTSAM
1490 NEXT M
1500 STRAIN.FEED! = TOTSAM/16
1510 STRAIN.OFF! = STRAIN.DES! - STRAIN.FEED!
1520 IF PROG=3 THEN GOTO 1540
1530 \* IF ABS(STRAIN.OFF!) >= (STRAIN.INC.DIGIT! * 3) THEN GOTO 2400
1540 IF ABS(STRAIN.OFF!) < = .5 THEN TOTSAM=0 : STRAIN.SAMP%=0 : GOTO 1610
1550 FINAL.STRAIN! = FINAL.STRAIN! + STRAIN.OFF!
1560 FINAL.STRAIN% = FINAL.STRAIN!
1570 IF FINAL.STRAIN% < INITIAL.STRAIN% THEN INC = -1
1580 TOTSAM = 0 : STRAIN.SAMP% = 0
1590 GOTO 1320
1600 \* 
1610 TIM.LOOP = ((TIMER + TIM.INC)-(INT((TIMER + TIM.INC)/TIM.INC) * TIM.INC))
1620 IF TIM.LOOP > = 2 THEN GOTO 1650
1630 T(F) = TIMER
1640 IF (T(F) - T(F-1)) < = 2! THEN F = F + 1 : GOTO 1460 ELSE F = 1:

Appendix A.
GOTO 1750
1650 C$ = INKEY$
1660 IF C$ = "E" OR C$ = "e" THEN GOTO 2400
1670 F = 1
1680 IF Z < > 0 THEN GOTO 1460
1690 L = L - 1
1700 IF L = 0 THEN GOTO 1750
1710 GOTO 1460
1720 ' Routine to sample the strain and stress
1730 ' 1750 IF PROG < > 3 THEN GOTO 1780
1760 IF Q < > 0 THEN Q = Q - 1 ELSE Q = ((120 / TIM.INC) - 1) + 1
1770 ' 1780 FOR J = 1 TO 16
1790 CALL ADC.VALUE (CHANEL.ZERO%, GAIN.ONE%, STRAIN.SAMP%)
1800 CALL ADC.VALUE (CHANEL.ONE%, GAIN.ONE%, LOADS.SAMP%)
1810 TOTSTR = STRAIN.SAMP% + TOTSTR
1820 TOTLOA = LOADS.SAMP% + TOTLOA
1830 NEXT J
1840 IF PROG = 3 THEN SWAP R, I
1850 TIM$(I) = TIME$
1860 IF AE = 0 THEN GOTO 1880
1870 AERMS(I) = ((TOTAE / CT) - 2048) * LSB.AE!
1880 IF PROG = 3 THEN SWAP R, I
1890 ' Data manipulation
1900 ' 1910 ' 1920 STRAIN.DIGIT! = TOTSTR / 16
1930 LOADS.DIGIT! = TOTLOA / 16
1940 IF PROG = 3 THEN GOTO 1980
1950 IF ABS(LOADS.DIGIT! - 2048) < = 4 THEN GOTO 2510
1960 STRAIN.VOLTS! = (STRAIN.DIGIT! * LSB!) + LOW.V!
1970 LOADS.VOLTS! = (LOADS.DIGIT! * LSB!) + LOW.V!
1980 IF PROG = 3 THEN SWAP R, I
1990 STRAIN.RECVD(I) = STRAIN.VOLTS! * STRAIN.PER.VOLT!
2000 LOADS.RECVD(I) = LOADS.VOLTS! * LOADS.PER.VOLT!
2010 PRINT USING "&"; TIM$(I),
2020 IF AE = 0 THEN GOTO 2090
2030 PRINT USING "#####"; STRAIN.RECVD(I) * 1000000!; LOADS.RECVD(I);
2040 E(I) = (LOADS.RECVD(I) / CROSS.AREA!) / (STRAIN.RECVD(I) * 1000000!)
2050 PRINT USING "#####"; E(I);
2060 PRINT USING "#####"; E(I) - E(I - 1);
2070 PRINT USING "#####"; AERMS(I)
2080 GOTO 2130
2090 PRINT USING "#####"; STRAIN.RECVD(I) * 1000000!;
2100 LOADS.RECVD(I);
2110 E(I) = (LOADS.RECVD(I) / CROSS.AREA!) / (STRAIN.RECVD(I) * 1000000!)
2120 PRINT USING "#####"; E(I);
2130 GOTO 2140
2140 IF D$ = "N" OR D$ = "n" THEN GOTO 2210
2150 OPEN E$ FOR APPEND AS #1
2160 IF AE = 0 THEN GOTO 2190
2170 WRITE #1, TIM$(I), STRAIN.RECVD(I), LOADS.RECVD(I), E(I), E(I) - E(I - 1),

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AERMS(I)

2180 GOTO 2200
2190 WRITE #1,TIMS(I),STRAIN.RECVD(I),LOADS.RECVD(I),E(I),E(I)-E(I-1)
2200 CLOSE #1

2210 TOTSTR = 0 : TOTLOA = 0 : TOTAE = 0 : CT = 0
2220 STRAIN.SAMP% = 0 : LOADS.SAMP% = 0 : AERMS.SAMP% = 0
2230 I = 1 + 1
2240 IF PROG=4 THEN IF I=7 THEN ITER%=368
2250 IF PROG=2 THEN GOTO 2270
2260 IF (ITER% + (I-1)) MOD ITER% = 0 THEN PRINT : GOTO 1200 ELSE GOTO 2350

2270 ITER.INC% = ITER.INC% + 1
2280 COUNT% = ITER.INC% MOD ITER%
2290 IF COUNT% < = 1 THEN GOTO 2340
2300 DELTE = E(I·2) - E(|·1)
2310 STRAIN.MUS.DES = (STRAIN.DES.VOLT! + LOW.V!) ' STRAIN.PER.VOLT!
2320 DROPE = (LOADS.RECVD(I-1) ' .0005)/(STRAIN.MUS.DES * CROSS.AREA!*1000000)
2330 IF DELTE < DROPE THEN ITER.INC% = ITER.INC% - COUNT% : PRINT : GOTO 1200
2340 IF COUNT% = 0 THEN PRINT : GOTO 1200
2350 Z=1:F= F+1
2360 GOTO 1460

2370 ' Routine to ramp down to 10 microstrain and end
2390 '
2400 FOR STRAIN.TO.APPLY% = FINAL.STRAIN% TO 2048 STEP -1
2410 CALL DAC.VALUE (DAC.SELECT%, STRAIN.TO.APPLY%)
2420 CALL ADC.VALUE (CHANEL.ONE%,GAIN.ONE%,LOADS.SAMP%)
2430 LOW.LOAD! = (LOADS.SAMP% * LSB!) + LOW.V!
2440 IF LOW.LOAD! < .2 THEN GOTO 2470
2450 FOR K = 1 TO 10: NEXT K
2460 NEXT STRAIN.TO.APPLY%
2470 IF CS="E" OR CS="e" THEN PRINT" TEST HAS BEEN PURPOSELY ENDED AT "; TIMES :GOTO 2550
2480 IF PROG=3 THEN GOTO 2500
2490 IF ABS(STRAIN.OFF!) >= (STRAIN.INC.DIGIT! * 3) THEN PRINT "EXTENSOMETER HAS MALFUNCTIONED : THE TEST HAS ENDED AT ";TIMES : GOTO 2550
2500 PRINT " STRAIN HAS REACHED DESTINATION : THE TEST HAS ENDED AT " ;TIMES :GOTO 2550
2510 PRINT " HYDRAULICS SHUT OFF NEAR ";TIMES;" : TEST HAS ENDED"
2520 ' Routine to recollect and print data
2530 ' Routine to recollect and print data
2540 ' 2550 SOUND 800,36
2560 IF D$="N" OR D$="n" THEN GOTO 2570 ELSE CLOSE
2570 PRINT; INPUT "Do you wish to print out the data collected (Y/N) ? ";F$
2580 IF F$="N" OR F$="n" THEN IF PROG=3 THEN GOTO 2810 ELSE GOTO 2800
2590 PRINT; PRINT "TO COLLECT AND PRINT THE DATA, FIRST INSURE THAT THE PRINTER IS ON AND READY:"; PRINT;PRINT," THEN, HIT %Ctrl-End"":
2600 KB$= INKEY$ : IF LEN(KB$)=2 THEN GOTO 2810 ELSE GOTO 2800
2610 IF RIGHT$(KB$,1) = "u" THEN GOTO 2820 ELSE GOTO 2860
2620 IF AE=0 THEN GOTO 2650
2630 LPRINT " TIME STRAIN LOAD MODULUS DMOD"
2640 GOTO 2660
2650 LPRINT " TIME STRAIN LOAD MODULUS DMOD"
2660 IF PROG=3 THEN SWAP R, I: I=I+1
2670 FOR P= 1 TO (I-1)
2680 LPRINT USING "&";TIM$(P),
2690 IF AE=0 THEN GOTO 2750
2700 LPRINT USING "#####";STRAIN.RECVD(P)*1000000I;LOADS.RECVD(P);E(P);E(P)·E(P-1);AERMS(P)
2710 GOTO 2780
2720 LPRINT USING "#####";STRAIN.RECVD(P)*1000000I;LOADS.RECVD(P);E(P);
2730 LPRINT USING "#####";STRAIN.RECVD(P)*1000000I;LOADS.RECVD(P);E(P)-E(P-1)
2740 NEXT P
2790 IF PROG=3 THEN I= I-1 : SWAP R, I : GOTO 2810
2800 END
2810 PRINT
2820 INPUT "Do you wish to store the data on a disk (Y/N) ? ",G$
2830 IF G$="N" OR G$="n" THEN GOTO 3020
2840 PRINT
2850 INPUT "Input the filename (disk:filename.filetype)—",E$ :
2860 E$= "A:DATERR.DAT"
2870 OPEN E$ FOR OUTPUT AS #1
2880 FOR Y= 1 TO R
2890 WRITE #1," TIME STRAIN LOAD MODULUS DMOD"
2900 GOTO 2920
2910 WRITE #1," TIME STRAIN LOAD MODULUS DMOD"
2920 CLOSE #1
2930 OPEN E$ FOR APPEND AS #1
2940 FOR Y= 1 TO R
2950 IF AE=0 THEN GOTO 2980
2960 WRITE #1,TIM$(Y),STRAIN.RECVD(Y),LOADS.RECVD(Y),E(Y),E(Y)-E(Y-1),AERMS(Y)
2970 GOTO 2990
2980 WRITE #1,TIM$(Y),STRAIN.RECVD(Y),LOADS.RECVD(Y),E(Y),E(Y)-E(Y-1)
2990 NEXT Y
3000 CLOSE #1
3010 ON ERROR GOTO 0
3020 GOTO 2800

AE Data Manipulation

DIMENSION RMSL(20),RMSS(2000),AE(2000),P(20)
C ENTER THE AE DATA OBTAINED FROM THE AE
DATA (RMSL(M),M = 1,8) /0.124609,0.123201,0.120409,0.116964,0.11424

Appendix A.
14.0,114244,0.114244,0.112225/
DATA (P(K),K=1,8) /8*0.0/
READ(7,*)NPTS
INC = 21
I = 1
DO 100 J = 1,NPTS
READ(7,*)|E,IE,IE,XJ,XL,XMOD,YMOD,AE(J)
C MANIPULATE TO REFLECT A POWER TERM
RMSS(J) = (AE(J)**2)*RMSL(I)
P(I) = P(I) + RMSS(J)
L = J + INC
LR = MOD(L,INC)
IF (LR.NE.0) GOTO 100
WRITE(8,2)P(I)
I = I + 1
100 CONTINUE
2 FORMAT(1X,1PE10.3)
STOP
END
Appendix B.

According to classical lamination theory (CLT), the relationship between the stress result-ant, \( N \), and the laminate strain, \( \varepsilon \), is expressed by:

\[
\{N\} = [A] \{\varepsilon\},
\]  
\[(C1)\]

where, \([A]\) is the in-plane, laminate stiffness matrix.

Equation (C1) may be inverted so that:

\[
\{\varepsilon\} = [A]^{-1} \{N\}.
\]  
\[(C2)\]

Let \( a_{ij} \) be the components of \([A]^{-1}\). If the load is applied uniaxially, i.e.,

\[
\{N\} = \{N_1 \ 0 \ 0\},
\]  
\[(C3)\]

then, from (C2), \( \varepsilon_{11} \) becomes:

\[
\varepsilon_{11} = a_{11} N_1.
\]  
\[(C4)\]

Dividing through by \( a_{11} \):

\[
N_1 = \frac{1}{a_{11}} \varepsilon_1.
\]  
\[(C5)\]
Recognize that the term \( \frac{1}{a_{11}} \) may be found from \([A]^{-1}\), or:

\[
\frac{1}{a_{11}} = \frac{A_{11}A_{22}a_{66} - A_{12}a_{66}}{A_{22}a_{66}} = A_{11} - \frac{A_{12}^2}{A_{22}} \tag{C6}
\]

where, of course, \( A_{ij} \) are the components of \([A]\).

In a quasi-isotropic laminate, \( A_{11} = A_{22} \). Substituting this into (C6) and that result into (C5) gives:

\[
N_1 = \frac{A_{11}^2 - A_{12}^2}{A_{11}} \varepsilon_1 \tag{C7}
\]

Dividing each side of (C7) by the laminate thickness, \( t \), yields:

\[
\sigma_1 = \frac{A_{11}^2 - A_{12}^2}{A_{11}t} \varepsilon_1 \tag{C8}
\]

Knowing that

\[
\sigma = E\varepsilon \tag{C9}
\]

then, from (C8),

\[
\bar{E}_1 = \frac{A_{11}^2 - A_{12}^2}{A_{11}t} \tag{C10}
\]

where, \( \bar{E}_1 \) is the effective laminate stiffness in the load-direction (signified as the "1" direction, for the sake of simplicity).

The components \( A_{11} \) and \( A_{12} \) may be expressed in terms of the invariants of the ply moduli [75]. For a quasi-isotropic laminate, from [75],

\[
A_{11} = U_1 t \tag{C11}
\]
where, \( t \) is, again, the thickness of the laminate and \( U_1 \) is an invariant expressed as a function of the principal ply moduli by:

\[
U_1 = \frac{1}{8} [3Q_{xx} + 3Q_{yy} + 2Q_{xy} + 4Q_{ss}] .
\]  

(C12)

In addition, \( A_{12} \) is related to another ply invariant, \( U_4 \), by the equation:

\[
A_{12} = U_4 t ,
\]  

(C13)

where, \( U_4 \) is given as:

\[
U_4 = \frac{1}{8} [Q_{xx} + Q_{yy} + 6Q_{xy} - 4Q_{ss}] .
\]  

(C14)

In a quasi-isotropic laminate the Poisson ratio, \( v_{xy} \), may be expressed in terms of the laminate stiffness components:

\[
v_{xy} = \frac{A_{12}}{A_{11}} ,
\]  

(C15)

or, from (C11) and (C13):

\[
v_{xy} = \frac{U_4}{U_1} .
\]  

(C16)

For the baseline (non-interlayered), quasi-isotropic laminate, the effective laminate modulus, \( \bar{E}_b \) may be found from (C10), (C11), (C13), and (C16):

\[
\bar{E}_b = \frac{(U_1t_b)^2 - (v_{bs}U_1t_b)^2}{(U_1t_b)t_b} ,
\]  

(C17)

where, \( t_b \) is the thickness of the baseline laminate and \( v_{bs} \) is the Poisson ratio of the baseline laminate.

Obviously, (C17) may be reduced to:

\[
\text{Appendix B.}
\]  

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\[ E_b = U_1(1 - \nu_b^2) , \quad (C18) \]

An interlayered, quasi-isotropic laminate is simply a baseline, quasi-isotropic laminate with the addition of several isotropic layers. Therefore, the laminate stiffness in the load-direction for the interlayered, quasi-isotropic laminate would be:

\[ A_{11} = U_1 t_b + U_1' (t_i - t_b), \quad (C19) \]

where, \( t_i \) is the total thickness of the interlayered laminate, \( t_i - t_b \) is the total thickness of the interlayer material, and \( U_1' \) is the first invariant of the interlayer material.

In the same manner,

\[ A_{12} = U_4 t_b + U_4' (t_i - t_b), \quad (C20) \]

or, by employing (C18):

\[ A_{12} = \nu_b U_1 t_b + \nu_i U_1' (t_i - t_b), \quad (C21) \]

where, \( \nu_i \) is the Poisson ratio of the interlayer material itself.

Substituting (C19) and (C21) into (C10) yields:

\[ E = \frac{(U_1 t_b + U_1' t_b)^2 - (\nu_b U_1 t_b + \nu_i U_1' t_b)^2}{(r t_b + t_b)(U_1 t_b + U_1' t_b)}, \quad (C22) \]

where, \( r \) is defined as:

\[ r = \frac{t_i - t_b}{t_b}. \quad (C23) \]

The ratio of \( \frac{E_i}{E_b} \) is sought. From (C22) and (C18) and some mathematical manipulation, this becomes:

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\[
\frac{\bar{E}_i}{E_b} = \frac{U_1^{2}t_0^2(1-v_b^2) + U_1^{2}r^2t_0^2(1-v_i^2) + 2U_1U_1^{'}r^2t_0^2(1-v_b^2)}{U_1(1-v_b^2)(U_1t_0 + U_1^{'}rt_0)(rt_b + t_0)}. \tag{C24}
\]

The first ply invariant of the interlayer material, \(U_1^{'}\), may be considered a constant fraction, \(f\), of the first ply invariant of the baseline laminate, \(U_1\), or:

\[
U_1^{'} = fU_1. \tag{C25}
\]

Plugging (C25) into (C24) and simplifying, results in:

\[
\frac{\bar{E}_i}{E_b} = \frac{(1-v_b^2) + f^2r^2(1-v_i^2) + 2fr(1-v_b^2)}{(1-v_b^2)(1+fr)(1+r)}. \tag{C26}
\]

A \((1 - v_i^2)\) is factored out of the numerator and denominator on the right-hand side of (C26), leaving:

\[
\frac{\bar{E}_i}{E_b} = \frac{1 + f^2r^2}{(1-v_b^2)(1+fr)(1+r)}. \tag{C27}
\]

A major simplifying assumption is made here. The contention is that the two ratios in the numerator are approximately equal to 1. Doubt is cast on this assumption if \(v_i\) is quite different than \(v_b\). Utilizing the assumption results in:

\[
\frac{\bar{E}_i}{E_b} \cong \frac{1 + fr}{1 + r}. \tag{C28}
\]

Incorporating (C25) into (C28) yields:

\[
\frac{\bar{E}_i}{E_b} \cong \frac{1 + \frac{U_1^{'}}{U_1}r}{1 + r}. \tag{C29}
\]

where \(r\) is defined in (C23).

Appendix B.
The vita has been removed from the scanned document