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Sotiris Kellas and John Morton
Virginia Polytechnic Institute and State University
Blacksburg, Virginia

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SCALING EFFECTS IN ANGLE-PLY LAMINATES

Sotiris Kellas
and
John Morton

Department of Engineering Science and Mechanics
Virginia Polytechnic Institute and State University
Blacksburg, VA 24061 - 0219

ABSTRACT

The effect of specimen size upon the response and strength of ±45° angle-ply laminates has been investigated for two graphite fiber reinforced plastic systems and several stacking sequences. The first material system was a brittle epoxy based system, AS4 fibers in 3502 epoxy, and the second was a tough thermoplastic based system, AS4 fibers in PEEK matrix. For the epoxy based system two generic ±45° lay-ups were studied; $(+45°_n/-45°_n)_{2S}$ (blocked plies), and $(+45°/-45°)_{2nS}$ (distributed plies), where n=1, 2, 3 and 4. In the case of the thermoplastic system only the lay-up with distributed plies was investigated; $(+45°/-45°)_{2nS}$, for n=1, and 2. The in-plane dimensions of the specimens were varied such that the width/length relationship was 12.7xn /127xn mm, for n=1, 2, 3, or 4.

It is shown that the stress/strain response and the ultimate strength of these angle-ply laminates depends on the laminate thickness and the type of generic lay-up (whether the plies are blocked or distributed). The ultimate strength of the epoxy matrix material was found to be much more sensitive to specimen size when compared to the thermoplastic matrix system. Scaling effects defined with respect to the first ply failure, strain at ultimate failure, and ultimate strength are isolated and discussed. Furthermore, it is shown that first ply failure occurs in the surface plies as a result of normal rather than shear stresses. The implications of the experimental findings upon the validity of the ±45° tensile test which is used to determine the in-plane shear response of unidirectional composites are discussed.

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1. INTRODUCTION

It is well known that some engineering materials exhibit strength scaling effects; that is, the strength of the material is some function of its absolute size, as well as a function of its geometric features. In particular, it has been observed that brittle isotropic materials which are subjected to tensile loads exhibit strength scaling effects which can be described by one of two scaling relationships. The first relationship, Eq.1, is based on Weibull statistics and states that the strength ratio, S_1/S_2, of two geometrically similar specimens is related to the ratio of the material volume, V_2/V_1, as:

$$\frac{S_1}{S_2} = \left(\frac{V_2}{V_1}\right)^{\frac{1}{m}} \tag{1}$$

where, $V_2 > V_1$ and m is known as the shape parameter, thought to be constant for a given material [1]. The second relationship, Eq.2, is based on a fracture mechanics approach, [2], and states that the ratio of critical stresses, σ_1^c/σ_2^c, is related to a scaling factor λ as:

$$\frac{\sigma_1^c}{\sigma_2^c} = \frac{1}{\sqrt{\lambda}} \tag{2}$$

where the critical stress, σ^c, is defined within the framework of linear elastic fracture mechanics as the stress required for unstable crack propagation, and λ is generally defined as the ratio of a geometric size in the model to that of the full scale structure.

In the case of unidirectional composites, under tensile loads, it has been shown that the Weibull statistics based model can be used to describe the changes in strength with specimen size within reasonable limits [3]. However, composites of practical importance are usually laminated with off-axis ply orientations and simple scaling laws, or models, become inadequate in the description of the strength scaling effects. Laminated composites are complex structures and their structural response bears no simple relationship to the size of the individual unidirectional plies. Moreover, in the case of laminated composites, the definition of a scaling effect is rather more complex since first ply failure can be as important as stiffness, strain at failure, and strength. In other words, two scaled specimens of the same lay-up may exhibit similar strengths even though first ply failure may occur at different applied stresses. In this simple example, the scaling effect from the ultimate strength point of view is negligible. However, a scaling effect does exists for the first ply failure and, under certain design requirements, this may be the most important consideration.

The in-plane geometry of laminated composite materials can be scaled by simply varying the in-plane dimensions according to some known ratio, λ, called the geometric scale factor. However, since the fiber diameter cannot be scaled there are two practical methods of scaling the laminate thickness. The first method is known as "ply level scaling" in which case a full scale laminate is constructed by increasing the individual ply thickness by blocking several layers of identical fiber orientation [4-6]. The second method is known as "sublaminate level scaling", where a full scale laminate is constructed by repeating a basic sublaminate about the plane of symmetry. For angle ply laminates where the $(+45°/-45°)_{2S}$ laminate represents the baseline, or model stacking sequence, $(+45°_n/-45°_n)_{2S}$ and $(+45°/-45°)_{2nS}$ represent the ply and sublaminate level scale model laminates, respectively. The two basic methods of scaling, for n=2, are also represented schematically in figure 1. If the in-plane dimensions of these scale model laminates are also sized in proportion to the model, the scaling procedure is known as three dimensional. If the thickness of the model and full scale laminates is the same but the in-plane dimensions are scaled the procedure is referred to as in-plane scaling, or two dimensional scaling. On the other hand, the in-plane dimensions can remain fixed for the model and the full scale laminates. In this case, the scaling procedure is referred to as thickness scaling, or one dimensional scaling. All three methods of dimensional scaling are shown schematically in figure 2.

Even though some three dimensional scaling studies have been performed [4-6], the bulk of the published research has dealt with one or two dimensional scaling problems [7-10], due to economic and experimental constraints. In the present experimental study, the influence of specimen size upon the stress/strain response of ±45° laminates is investigated for cases of one, two and three dimensional scaling. In addition, both ply and sublaminate thickness scaling levels are studied. While the largest portion of this experimental study is centered around the behavior of a graphite/epoxy material system, a second system, graphite/PEEK, was introduced, and studied briefly, to provide a better understanding of the problem of scaling in matrix dominated lay-ups. The purpose of choosing the ±45° lay-up for this investigation is threefold. Firstly, this represents the only matrix dominated lay-up where, under a uniaxial load, the plies are loaded primarily in shear. Secondly, surface ±45° plies are a necessary constituent element in most structural laminates where damage tolerance is a requirement. Finally, the ±45° tensile test is one of the most popular tests for determining the shear stress/strain response of unidirectional composites. Consequently, any dependencies of the stress/strain response, first ply failure, and strength on specimen size and stacking sequence should be well understood and characterized.

2. EXPERIMENTAL DETAILS

2.1 Material Systems

Two material systems with identical fibers were studied; AS4 fibers in 3502 epoxy matrix and AS4 fibers in PEEK thermoplastic matrix. The graphite/epoxy panels were layed up and cured in-house according to manufacturer's specifications. The graphite/PEEK panels were obtained in their moulded form, directly from the manufacturer.

2.2 Stacking Sequence

Two generic ±45° lay-ups were studied, one with blocked plies and one with distributed plies with stacking sequences ranging from 8 to 32 plies. The 8-ply laminate consisted of unidirectional plies arranged in a $(+45°/-45°)_{2S}$ sequence, and was denoted the baseline or model stacking sequence. For the epoxy based system, six additional "scaled-up" laminates were tested with the following stacking sequences; $(+45°_n/-45°_n)_{2S}$ (blocked plies) and $(+45°/-45°)_{2nS}$ (distributed plies), where n = 2, 3, and 4. For the PEEK matrix system, one "scaled-up" laminate was tested with stacking sequence; $(+45°/-45°)_{4S}$ (distributed plies), in addition to the baseline stacking sequence.

2.3 Specimen Geometry

For three dimensional scaling (defined in figure 2), the general specimen dimensions were 12.7xn mm wide, 127xn mm long and 1.0xn mm thick, where n=1, 2, 3 and 4. For one and two dimensional scaling, the same general dimensions were used with either the thickness or in-plane dimensions being held constant and n varied from 1-3. The full test matrix is shown in tables 1 and 2 for the epoxy and the PEEK matrix materials, respectively.

2.4 Loading Mode

All tests were carried out in tension at a constant strain rate equal to 0.028 min^{-1}. In the early part of the program [6], the strain was monitored with the aid of extensometers and strain gages. However, due to the limited capability of strain gages in measuring large strains, in specimens with surface cracks, subsequent tests on specimens with distributed plies were performed with extensometers only. Both custom built extensometers, which were designed to measure strains over scaled gage areas, and commercial extensometers with a fixed gage section (25.4 mm gage-length) were used. Four custom-built extensometers with scaled gage lengths of 38xk mm, where k=1, 2, 3, and 4, for specimen size **a**, **b**, **c**, and **d** respectively. The gage length for the custom-built extensometers was defined by means of two parallel grooves. Each groove was scribed into a thin layer of soft epoxy (approximately 12 mm wide) which was brushed on the surface of each specimen, before testing. This technique was adopted following previous tests [6], which indicated that slipping of the extensometer knife edges occurs. Such slipping was evident only in laminates with off-axis surface plies and appeared to be compatible with fiber in-plane rotation following an accumulation of matrix transverse cracks throughout the specimen volume. The transverse strains were not measured.

2.5 Non-Destructive Damage Evaluation

Non-destructive damage examination was carried out on specimens pre-loaded to a given stress level. The given stress levels for each specimen type were selected according to the predetermined stress/strain curves for a given family of specimens. Following the first loading/unloading cycle, specimens were soaked into a zinc iodide solution before being X-rayed. Subsequently both free edges of loaded specimens were carefully polished and then examined under an optical microscope. In most cases the whole procedure was repeated again with a load step slightly higher than the preceding one. At least one specimen of each specimen type was examined in this manner.

3. EXPERIMENTAL RESULTS

3.1 Stress/Strain Response

Typical stress/strain plots of specimens scaled in three dimensions are shown in figures 3 and 4 for the epoxy and the PEEK matrix systems, respectively. The plots shown in these figures are typical of a given specimen type from a number of replicate tests (at least seven replicate tests were performed for each specimen type and size).

In the case of the epoxy matrix system, typical stress/strain plots are presented in figure 3; for the 8-ply baseline case and for each generic lay-up (distributed or blocked plies), with n = 2, 3 and 4. The stress/strain response of the ±45° laminates depends on the specimen size and the stacking sequence. Specimens with blocked plies exhibit a brittle-like behavior, whereas specimens with distributed plies exhibit a ductile-like behavior. Moreover, specimens with blocked plies show a consistent reduction in strength with increasing specimen size. To the contrary, an increase in strength with increasing specimen size is exhibited by the specimens with distributed plies. Note that, while the strain at failure for the 32-ply specimen with blocked plies is approximately 1%, the strain at failure of the corresponding 32-ply specimen with distributed plies has exceeded 12%. Furthermore, for these two specimens the difference in the ultimate strength is approximately 170%. Clearly, so far as the ultimate strength and strain at failure are concerned, there are significant scaling effects. The magnitude and direction of the scaling effect whether increasing or decreasing relative to the baseline, depends upon whether the laminate was scaled on a ply or sublaminate level.

The average values of ultimate stress and strain at failure, for the epoxy matrix specimens scaled at the sublaminate level, are presented in tables 3 and 4, respectively. Note that the coefficient of variation, which is shown in brackets, in tables 3 and 4 is consistently higher for the strain at failure measurements when compared to the corresponding values of the ultimate strength. Moreover, for the same specimen types, the strains measured with the custom built extensometers were always slightly higher than those measured with the MTS extensometer, as indicated in table 4. Since the knife edges of the custom built extensometers were secured in grooves, knife edge slipping which is associated with fiber scissoring was eliminated.

In general, epoxy matrix specimens scaled at the sublaminate level with 24 plies or more exhibited two peaks in their stress/strain curves; for example see specimens **Cc** and **Dd** in figure 3. It is thought that the apparent increase in stiffness, or strain hardening, following the first peak stress, results from the alignment of the fibers with the loading direction. This phenomenon will be referred to as fiber scissoring. The average of the stress and strain values corresponding to the first peak, which are thought to be more relevant to the determination of the shear strength, [11], were extracted from each individual plot, for specimens with 24 or 32 plies, and are presented in tables 5 and 6. Once more it should be pointed out that the coefficient of variation for the measured ultimate stresses was much lower than that for the corresponding strains at failure.

The stress/strain plots for AS4/PEEK specimens, shown in figure 4, also indicate the existence of scaling effects for the two specimen sizes **Aa** and **Bb**. As in the case of the epoxy matrix material for specimens with distributed plies, both the strength and the strain at failure increase with specimen size. However, when corresponding specimens are compared, there exists a basic difference between the stress/strain response of the two material systems. Unlike the epoxy matrix specimens, the PEEK matrix specimens do not exhibit a negative gradient in their stress/strain responses. Moreover, the PEEK matrix specimens appear to be more ductile when compared to the corresponding epoxy matrix specimens with a difference in strain at failure of approximately 15% strain, as depicted in figure 5, which shows a comparison of the stress/strain response of a baseline epoxy and a baseline PEEK matrix specimen. In the magnified portion of figure 5, it is also shown that the initial extensional stiffness of the PEEK matrix specimens is slightly lower than that of the epoxy matrix specimens.

The effect of laminate thickness for epoxy matrix specimens of the same in-plane dimensions is shown in figure 6. Clearly, the laminate thickness has a significant influence upon the stress/strain behavior. Specimens with distributed plies show an increase in strength and strain at failure with increasing laminate thickness, while the contrary is true for specimens with blocked plies. An almost identical effect was observed for the three dimensional scaled specimens, if the results of figures 3 and 6 are compared. This observation indicates that in-plane scaling will have a lesser effect on strength and strain at failure than thickness scaling. Indeed, the stress/strain plots in figure 7 show that, within a large portion of their stress/strain curve, epoxy matrix specimens with scaled in-plane dimensions exhibit a very similar response. The fact that narrow specimens (with width/thickness ratios less than 12.7) fail at somewhat lower strains than the corresponding wider specimens may be attributed to the higher gripping constraint rather than the stress field being different in the gage section of the specimens. This is supported by the observation that ultimate failure in the narrow specimens occurred either in or near the gripping region.

Comparisons of one and two dimensional scaled PEEK matrix specimens are shown in figure 8. While evidence of a definite dependency of strength upon the thickness dimension is not as strong as in the case of the epoxy matrix specimens, it is still clear that the laminate thickness is the predominant scaling dimension. Replicates of 8-ply PEEK matrix specimens showed a large scatter for strains greater than 1%, as shown in figures 9 and 10 for specimen sizes **a** and **b**, respectively. Note that there is a clear correlation between the stress/strain response of the 8-ply PEEK matrix specimens and the position within the panel from which the specimens were machined. In general, specimens closer to the edges exhibited a lower ultimate stress and strain at failure. However, the initial part of the stress/strain response of these specimens appears to be independent of the specimen position within the panel.

A comparison between 8 and 16 ply PEEK matrix specimens is shown in figure 11 where the ultimate stress and the strains at failure are plotted against the specimen position from the panel edge. Clearly, in the case of the 8-ply specimens there is a strong dependency between ultimate stress and specimen location, with differences in ultimate stress between the edge and the mid-panel specimens of 50%. Despite the obvious trends in ultimate stress, for the 8-ply specimens, no obvious trends for strains at failure, have been observed. This irregular pattern of the strains at failure, for the 8 ply specimens is thought to be an artifact of the method of strain measurement, rather than a true material behavior. Because of the smooth specimen surface finish, the method of attaching the extensometer knife edges to the specimens had to be varied, initially, until a satisfactory result was obtained. Most of these preliminary tests were carried out on 8-ply specimens, which can account for some of the inconsistencies, shown in figure 11. Moreover, unlike the ultimate stress values which were directly recorded, strains beyond 15% were extrapolated, using the tabulated stress values and an assumed constant rate of strain.

The average values of ultimate stress and strain at failure for PEEK matrix specimens are summarized in tables 7 and 8. Note that, as shown in figure 11, the ultimate stress was a function of the specimen position from the panel edges. Therefore, care must be exercised in interpreting the average values of ultimate stress in table 7. If, for example, the average of all seven **Aa** specimens, which were destructively tested, is compared to the average ultimate stress of the **Ab** specimens, it appears that the change in specimen in-plane dimensions has a considerable influence on the ultimate strength (a 24.7% increase from size **a** to **b**). However, if the average of specimens from an equivalent panel position is considered, it is clear that the influence of the in-plane specimen dimensions is not as important, since the average ultimate stress increases by only 3.5% from specimen sizes **a** to **b**, see table 7. As in the case of the epoxy matrix specimens, the thickness dimension appears to have a much greater influence on strength, than the in-plane dimensions. When fabrication process artifacts are eliminated, the results in table 7 show a 30.7% and 21.1% strength increase, for size **a** and **b** specimens respectively, when the thickness is increased from 8 to 16 plies.

3.2 Damage Evaluation

Thus far, scaling effects have been examined by considering stress/strain responses, ultimate strengths, and strains at failure. An additional important factor for laminated composites is the initiation of damage and first ply failure. Evidence of scaling effects related to first ply failure for the two material systems is presented in figures 12 and 13 for epoxy and PEEK matrix specimens, respectively. These figures show a combined view of damage as documented by enhanced X-ray radiographs of the gage section and micrographs of the polished specimen free edges. There is a clear correlation between damage initiation and stiffness changes observed in the stress/strain curve for a given specimen type.

In general, all ±45° laminates (with the exception of 24 and 32 blocked ply laminates) presented in figure 3 share approximately the same initial extensional stiffness. Changes in extensional stiffness are usually expected to occur when the material enters a non-linear state and/or when damage occurs. Clearly then, when a group of ±45° laminates follow an identical path up to a point where one deviates from the rest, then that point could coincide with failure initiation. For the 24 and 32 blocked ply laminates it has been shown [6] that damage exists in virgin (non-loaded) specimens. Therefore, the stiffness deviations for these two specimens from the rest, even though they are small at low strains, can be attributed to the observed transply cracks. In the case of distributed ply specimens, where no damage was observed in virgin specimens, noticeable stiffness deviations occur at approximately 150 MPa. The first specimen to show such a deviation is the baseline (8-ply) specimen. Damage examination of these specimens, pre-loaded to approximately 150 MPa, showed that damage has initiated, see figure 12a, and that it is confined to the surface plies. Further damage occurred in the middle plies at a stress of 158 MPa which corresponds approximately to the peak value of stress for this specimen, see figure 12(b).

Following the same argument, failure in the 8-ply PEEK matrix specimens should occur at approximately 85 MPa where the stress/strain curve of the 8-ply specimen deviates from that of the 16-ply specimen, as indicated in figure 4. Microscopic examination of polished edges of specimens loaded to 85 MPa and beyond, did not produce evidence of transply cracks. However, X-ray radiographs did contain dark lines at 45° which are normally associated with transply cracks. Damage, as detected in the X-ray radiographs, became clear in PEEK matrix specimens which were loaded to a stress of approximately 90 MPa and beyond. At this point there is no conclusive experimental evidence as to the nature of this damage, but one possible explanation is fiber debonding. Once several neighboring fibers debond from the supporting matrix, their effect on the stress/strain response of the laminate could be similar to that of transply cracking and, such damage is X-ray detectable. Clear evidence of transply cracking in the 8-ply PEEK matrix specimens was observed at stresses of 117 MPa and beyond, as shown in figure 13(a). Such cracks were first observed to occur in the surface plies (the +45° direction) and, as the X-ray radiograph of figure 13(a) indicates, there co-exists damage in the -45° direction. The damage in the -45° direction is, however, not seen on the photomicrograph of the specimen's edge. That is, while some of the dark lines at +45° correspond to transply cracks, a large number of lines at +45° and -45° may be associated with fiber debonding. Moreover, comparing figures 13(a) and 13(b) it can be seen that equivalent damage in the 16-ply specimens occurs at approximately 131 MPa: which is an increase of approximately 14% from the stress level at which damage was seen in the 8-ply laminates. Likewise, from the data in figure 3, it can also be deduced that first ply failure in the 16-ply epoxy specimens occurs at approximately 159 MPa, compared to 150 MPa for the 8-ply specimens, which represents a 6% increase in first-ply-stress. The delay in damage development from one specimen size to the next is a scaling effect, and this type of scaling effect appears to depend on the material system. For a direct comparison, the stress values at which damage initiates in the baseline epoxy and PEEK matrix specimens are indicated in figure 5.

Another important difference in the damage development in different size specimens is shown in figures 14(a) and 14(b) for two epoxy matrix specimens loaded to a stress corresponding to the point on the stress/strain curve where the gradient is zero, see figure 3. While damage in the 8-ply specimen develops around the initial damage, damage in the 16-ply specimen is more dispersed throughout the gage section, as seen in both the X-ray radiograph and the micrograph of such a specimen.

A comparison of the damage between ply and sublaminate level scaled specimens is shown in figure 15 for 32 ply thick specimens. Note that, at zero applied load the ply level scaled specimen is severely cracked in all plies, and, as indicated in the micrograph of the polished edge, the cracks appear to be related primarily with the opening mode.

4. DISCUSSION

Scaling effects in laminated composites can be defined with regard to several parameters, the most important of which are strength, strain at failure, stress/strain response, and first-ply-failure stress. If, for example, two geometrically scaled specimens exhibit different stress/strain responses this would be a scaling effect. Likewise, if the two specimens fail at different ultimate stresses this would also be a scaling effect. All four factors are important and these will be discussed separately, but, since the initiation of damage is the parameter that, in many cases, controls the strength and stress/strain response of a laminated composite, it will be discussed first. Finally, the effect of specimen size will be examined with reference to the standard ±45° tensile test, used to obtain the in-plane shear stress/strain response of unidirectional composite materials.

4.1 First Ply Failure

The understanding of the processes that control the first ply failure in angle-ply laminates has implications which extend beyond the demands of the present study, since most practical composite laminates contain angle-plies on their surfaces, for improved damage tolerance. Therefore, any factors that can influence the failure of these outer plies, such as the ply thickness, and constraint of the neighboring plies, should be identified and well characterized, so that, meaningful failure criteria can be established, which can account for size effects. Moreover, if the failure processes, in scaled

±45° laminates are well understood, more damage tolerant full scale composite structures can be designed.

From the experimental evidence it has been deduced that first ply failure occurs in the surface plies, followed by failure of the mid-plane blocked plies. This was found to be true for all specimen sizes and laminate thicknesses. Therefore, it appears that a ±45° laminate has two "weak links", the surface and the middle plies. The question is; what is the cause of weakness in these plies? If the classical lamination theory assumptions are applied, it is expected that all plies, in a symmetric ±45° laminate, would be strained in an identical manner, and therefore, first ply failure would occur arbitrarily in any one of the plies, not necessarily the surface plies. However, in practice, the top (free) surface of the outermost ply in a symmetric ±45° laminate is not constrained in the same way as the inner surface, which is bonded to an adjacent -45° ply. As a result, in-plane shear strains, γ_{xy}, develop on the surface, close to the free edge, of the laminate when an axial load is applied in the x-direction, as shown in figure 16. It can be shown, using a simple strain transformation approach that such shear strains can be transformed into significant normal strains which act transverse to the fiber direction, ε_2. Since the free surface material is unconstrained in the direction transverse to the fibers, cracking occurs. Therefore, first ply failure in ±45° laminates is a normal fracture (or fracture occurring in the opening mode), and not a shear mode fracture. This argument is supported by the opening nature and starting position of first ply cracks. An example of such a crack is shown in figure 17(a). Likewise, the mid-region of the center ...-45°/-45°... plies is subjected to a reduced constraint when compared to the interply region, where the constraint from the neighboring +45° plies is at a maximum. As a result, large normal strains can develop, in a similar manner but to a lesser degree than those in the surface plies, producing one more site for normal failure. The regions of high normal strain, ε_2, on the laminate free edge are shown schematically in figure 17.

The sequence of ply failures, for the epoxy matrix specimens, is shown in figures 12(a) & (b), where first ply failure in the surface ply occurs at a stress of 150 MPa, followed by second ply failures in the mid-plane region when the stress is equal to approximately 158 MPa. Moreover, figure 17(a) (which shows a close-up view of a surface transply crack) indicates that the crack has initiated at the junction between the specimen surface and the specimen free edge, and is propagating towards the first ply interface. Likewise, the close-up view of a crack in the mid-plane plies, figure 17(b), shows that the crack has initiated in the middle and is propagating towards the neighboring ply interfaces. Neither one of these cracks, shown in figure 17, is associated with interlaminar fracture (and, therefore, interlaminar stress concentrations), and both appear to propagate in an opening mode as a result of dominant normal stresses.

Since the ply constraint is a function of the ply thickness this appears to be an important scaling parameter. For thicker surface plies the constraint on the top surface is less and, therefore, the normal strain, ε_2, is greater. Likewise, the thicker the block of plies at the mid-plane of the laminate, the greater ε_2, and, therefore, the lower the first and second-ply-failure stresses. Following the manner in which first and second ply failures occur, the effect of these ply failures upon the different specimen types can be examined. For example, the first-ply-failure stress of specimens scaled at the ply level (figure 1), would be expected to be reduced as the ply thickness increases and, therefore, a lower ultimate strength and strain at failure would be expected. Indeed, as shown in figure 3, both the ultimate strength and the strain at failure are reduced with increasing specimen thickness. On the other hand, specimens scaled at the sublaminate level share the same surface ply thickness and the same mid-plane ply thickness; a single ply on the surface and two blocked plies in the middle. In this case the severity of the weak links is reduced with increasing specimen thickness and, therefore, the strength and strain at failure increase with increasing specimen thickness. Note that the $(+45°/-45°)_{2nS}$ laminate is equivalent to a $(+45°\{[-45°/+45°]_{2n-1}/-45°_2/[+45°/-45°]_{2n-1}\}+45°)$ laminate which, basically, is the same as (+45°{unbalanced core}+45°) where the shear/extension coupling of the unbalanced core will depend on n. The shear/extension coupling is reduced as n increases. Effectively, this means that the constraint at the interface between the unbalanced core and the surface plies is reduced as n increases, resulting in some degree of relaxation of the in-plane shear strain, γ_{xy}, on the free surfaces, and hence a higher first-ply-failure stress. The effect of the failed plies on the rest of the laminate also depends on the total number of plies, and this will be discussed below.

An issue concerning the importance of the specimen dimensions is raised in this investigation. Which is the most important specimen dimension? Is it the specimen thickness, or is it the specimen width, or is it the specimen width/thickness ratio? Figure 3 shows that the specimen size, for a given scale-up procedure, affects the first-ply-failure stress. For example, considering the specimens scaled at the sublaminate level, and looking at the points where the stress/strain curve of a given specimen size deviates from the rest, it can be deduced that the first-ply-failure stress for the 8 and 16 ply specimens is approximately 150 and 159 MPa, respectively. Figure 6(a), shows that even though the 16-ply specimen, **Ba**, has a narrower width (12.7 mm versus 25.4 mm in figure 3) than the 16-ply, **Bb** specimen, the first-ply-failure stress is still approximately 159 MPa. That is, the change in specimen width did not influence the first ply failure stress. Likewise, the 8-ply specimen, **Ab**, in figure 6(b) has a width twice as large as the **Aa** specimen of figure 3 and yet, first ply failure still occurs at approximately 150 MPa. Again, the specimen width did not affect the first ply strength. Clearly, the laminate thickness appears to be the most important specimen dimension. A remaining question is the importance of the specimen width/thickness ratio.

The issue of the specimen width/thickness ratio has been approached by many researchers who have anticipated that interlaminar stresses can influence the failure mechanisms of angle-ply laminates, e.g. [12]. Interlaminar stresses have also been thought to have been responsible for the different stress/strain responses in distributed and blocked ply ±45° laminates [13]. In some cases, interlaminar stresses were also thought to be important enough to influence the specimen design for the ±45° tensile specimen [14]. If such edge stresses were significant in the mechanical response of ±45° laminates, specimens of different width/thickness ratios would exhibit different first-ply-failure stresses. However, the present experimental findings indicate that, within the range of stresses that first ply failures occur, there is no significant difference in the stress/strain response of such laminates, as shown in figure 7. Clearly, specimens with the same number of plies exhibit a similar stress/strain response within a large portion of the stress/strain curve. In particular, the 24 ply specimens, **Ca** and **Cc** in figure 7(b), which have width/thickness ratios of 4.2 and 12.7, respectively, exhibit a more or less identical stress/strain response up to about 6% strain; well beyond the point of first ply failure. A similar behavior has also been observed in the PEEK matrix specimens, as indicated in figure 8. The fact that the narrow specimens fail at lower strains than the corresponding wider specimens is probably related to the higher grip constraint that the narrow specimens have to endure. For this reason, most narrow specimens failed either very close or within the gripping region.

When it was apparent that the first ply failure in the epoxy specimens occurred in a normal or a mixed normal/shear mode, rather than a pure shear mode, it was postulated that a tough matrix system with a high opening mode crack resistance would sustain higher loads prior to first ply failure. For this reason, the PEEK matrix specimens were introduced. However, as the experimental results indicate, the first-ply-failure stress for the PEEK matrix specimens appears to be substantially lower than the corresponding epoxy specimens, as shown in figure 5. Even if the first-ply-failure stress in the PEEK matrix specimen is considered to occur at the point where a visible (by optical microscope) ply crack initiates, that stress is still about 22% lower than the equivalent stress in the corresponding epoxy matrix specimen. Clearly, the toughness of the matrix alone is not a sufficient parameter for improved ply toughness. It would appear, at least from this investigation, that the fiber/matrix interfacial strength may be a more important factor. Additionally, another factor which contributes to the rather low value of first-ply-failure stress is the relative ply stiffnesses, which control the surface deformations of the laminate. This subject requires further investigation.

4.2 Stress/strain Response

Classical lamination theory predicts that undamaged ±45° laminates will exhibit the same initial stress/strain response regardless of the stacking sequence. This is clearly shown in figure 3, for the five laminates; **Aa, Bb, Cc, Dd** with distributed plies and **Bb** with blocked plies. The two laminates with blocked plies, **Cc** and **Dd** appear to have a slightly lower initial stiffness as a result of residual-stress induced damage in these specimens [6]. Clearly, the total stress/strain response of angle-ply laminates depends upon the generic lay-up (blocked or distributed plies) but also upon the laminate thickness. The stress/strain response of epoxy matrix specimens changes from brittle to ductile as the thickness of the blocked plies is reduced or the thickness of the distributed ply laminates is increased. These large changes in the stress/strain response can be attributed to transply cracking (or fiber/matrix debonding). When such cracks are formed, the laminate stiffness decreases initially, and, in the case of the epoxy specimens, the tangent stiffness eventually becomes negative. When the laminate stiffness reaches a minimum value the specimen either fails due to the fact that severe localized straining takes place in a small region of the gage section, in which the crack density is high; or, the specimen stiffness increases due to fiber alignment with the loading direction (this effect is known as fiber scissoring). In other words, if a $(+45°/-45°)_{2nS}$ laminate can survive a given deformation before ultimate failure occurs, the stiffness will increase as a result of fiber scissoring. Specimens that fall into this category are expected to be those with brittle matrices and a large number of ply interfaces like the epoxy matrix specimens with 24 or greater number of distributed plies, or specimens with a compliant matrix and/or a relatively weak fiber/matrix interface such as the PEEK matrix specimens. Once a significant amount of transply cracking occurs, the integrity of the laminate depends solely upon the ply interface region which holds the damaged plies together. Therefore, laminates with a large number of ply interfaces can endure larger axial strains. In this case, the uniform distribution of the transply cracks within the volume of the specimen is also important. The more evenly distributed the cracks, the higher the axial strain, which results in large fiber rotations which, in turn, results in a higher laminate stiffness and ultimate strength.

The absence of a negative gradient in the stress/strain response of PEEK matrix specimens constitutes a characteristic difference between corresponding PEEK and epoxy matrix specimens. The lack of a negative (or even zero) gradient in the stress/strain response of PEEK matrix specimens can be attributed to the larger strains that can be accommodated by the PEEK matrix, which promotes more fiber rotation (scissoring) for a given applied stress.

An observation which is characteristic of the 8-ply PEEK matrix specimens is the dependency of the stress/strain response on the specimen distance from the panel edges, as indicated in figures 9 and 10.

Since there is a strong correlation between the position of the specimen and the stress strain response, it is possible that the rate of cooling following fabrication, which can affect the material crystallinity, is responsible for the observed behavior. Since all specimens, in figure 9 or 10, share the same initial stress/strain response it would be reasonable to conclude that the shear stiffness is independent of a specimen distance from the panel edge. In other words, only the fracture characteristics are affected by the fabrication process. Moreover, the fact that the 8-ply specimens are much more sensitive to the position within the panel, when compared to the 16-ply specimens, indicates that the controlling factor is the strength of the surface plies in the direction transverse-to-the-fibers. Due to the nature of the loading mode the surface plies constitute the weakest link within a ±45° laminate. Consequently, premature failure of the weak surface plies in an 8-ply specimen is much more detrimental to the specimen performance than the failure of the same pair of plies in a 16-ply specimen.

4.3 Strength and Strain at Failure

It has been shown that the first ply failure, for a given generic ±45° lay-up (with blocked or distributed plies) depends primarily upon the laminate thickness. Moreover, it has been shown that, within a large range of applied loads, the stress/strain response depends primarily upon the laminate thickness. From the point of view of the ultimate strength and strain at failure, it appears that the laminate thickness is also the most important factor. Wide specimens, in general, exhibited a slightly higher ultimate strength and strain at failure; however, this was attributed to a loading artifact rather than an actual material scaling effect. This represents a typical practical constraint, where a loading condition, as simple as a tensile load, cannot be reproduced in an equivalent manner for any set of scaled specimens.

While, in the case of distributed ply laminates, the increase in first-ply-failure stress with specimen thickness is partly responsible for the observed increases in ultimate strength, the end result should depend on additional factors such as the number of ply interfaces within a laminate, the distribution of damage within the laminate, and the overall effect of the failed plies on the remainder of the laminate. For example, in an 8-ply laminate, failure of the surface plies followed by failure in the mid-plane plies, is roughly equivalent to a 50% ply loss. However, failure of the same four plies in a 32 ply thick laminate corresponds to only 12.5% ply loss. Clearly, the influence of the four failed plies will be detrimental in an 8-ply laminate. Likewise, in a 32-ply laminate, constructed out of blocked plies, the first and second ply failures correspond to a 50% ply loss (eight plies in the two +45° blocks on the surface and eight plies in the -45° block in the middle) and result in an equally detrimental effect on strength. Moreover, specimens with blocked plies exhibit a lower first-ply-failure stress. For some laminates, first ply failure occurs even before any mechanical load is applied to the specimens, due to the influence of curing stresses on the weaker (unconstrained), blocked plies [15]. As a result, the strength of the 32 ply laminate with blocked plies is much lower than the strength of the 8-ply baseline specimen.

For the purpose of a direct comparison, the normalized strength and strain at failure of the seven different types of ±45° epoxy matrix specimens, which were geometrically scaled in three dimensions, are plotted in figure 18. Note that for specimens scaled on the sublaminate level, very large strains were measured and fiber scissoring occurred. Consequently, at failure, the fibers are no longer oriented at ±45°. In fact, it can be shown, using a simple elasticity relationship, Eq.3, that, for a Poisson's ratio, $\nu_{xy} = 0.7$ (a typical value for a ±45°

$$\gamma_{12} = (1 + \nu_{xy})\, \varepsilon_x \tag{3}$$

graphite/epoxy laminate) and an applied strain $\varepsilon_x = 2\%$, the fiber rotation is approximately 1°. Therefore, an apparent ultimate strength of the ±45° laminates, scaled at the sublaminate level, was chosen to represent an average in-plane shear strength rather than using the laminate ultimate strength [11]. The in-plane shear stress is equal to a half of the applied stress, as given by Eq.4, using the analysis of Rosen [16].

$$\tau_{12} = P/2A \; (= \sigma_x/2) \tag{4}$$

The stress chosen to describe the apparent strength of the ±45° epoxy matrix laminates corresponds to the point where the gradient of the stress/strain curves first becomes zero, as shown in figure 3. Also, an apparent strain at failure corresponds to the same point on the stress/strain curve. This point for the 32-ply laminates corresponds to approximately 2.4% axial strain, a value small enough to neglect any fiber scissoring. As shown in figure 18, the apparent strength and strain at failure show the same trend with increasing specimen size: an increase for specimens with distributed plies and a decrease for specimens with blocked plies. Note that the strength and strain at failure of the specimens with distributed plies cannot show an indefinite increase in strength with specimen size. When a given large number of plies is reached, volume effects due to the variability of the material will become important and, therefore, a plateau in the curves of figure 18 is to be expected. In fact figure 18(b) indicates that the 32-ply laminate, with distributed plies, is very close to such a plateau.

4.4 The ±45° Standard Shear Test

An ideal shear test can be defined as one in which a known state of uniform and pure shear can be achieved. Even in the case of isotropic materials, this condition is difficult to attain and maintain for large deformations. For fiber reinforced composites the concept of a uniform stress state is relative and depends upon the scale of the constituent materials, the degree of anisotropy, and the laminate stacking sequence. In general, a state of uniform stress can exist only at the ply level and away from free edges. Consequently, errors are introduced in the shear property determination when, for example, the strain is measured from a strain gage applied to a surface ply, and an average value of the shear stress is used to calculate the shear stiffness. Inaccuracies in the shear strength occur when the ultimate shear load is divided by an average area when, in practice, failure occurs progressively after damage initiation at points of non-uniform stress.

There is a large number of test methods available for measuring the shear stiffness and strength of advanced composite materials. Each test method, however, represents a compromise between the ability to produce a reliable and/or meaningful result, and the total cost or complexity of the test procedure. The ±45° shear test is one of the simplest tests available for the determination of the in-plane shear stress/strain response of fiber reinforced unidirectional composites. In addition to its simplicity, the ±45° shear test, is one of few tests available that is used for the determination of both the shear stiffness and the shear strength. The test was originally proposed by Petit [17] and it was later improved (simplified) by Rosen [16]. Following Rosen's recommendations, the ±45° tensile test was adopted by the American Society for Testing and Materials (ASTM) as a standard test method, D-3518, in 1976. The effectiveness of this simple test was recognized by many organizations such as the Suppliers of Advanced Composite Materials Association (SACMA) and the Composites Research Advisory Group (CRAG). Consequently, the ±45° tensile test and the resulting data have gained wide popularity and acceptance in a variety of applications such as research and development, material selection and qualification, and structural design. Unfortunately, some of the assumptions involved in the test method for the determination of the shear strength, which were clearly stated by Petit [17] and Terry [13], are often overlooked. In particular, the effect of specimen geometry (size) upon the strength of the ±45° laminates has not been given sufficient attention.

The test procedure involves a balanced and symmetric ±45° composite specimen loaded in tension. The specimen dimensions are defined according to ASTM D 3039 which designates specimen thickness between 0.5 and 2.5 mm, that is, between 4 and 20 standard thickness (0.125 mm) unidirectional plies.

Using a simple equilibrium argument Rosen [16] showed that, for a state of uniform stress, the average in-plane shear stress can be obtained from Eq.4. In addition, using Eq.3, he expressed the in-plane shear strain in terms of the axial and transverse laminate strains, Eq.5.

$$\gamma_{12} = (\varepsilon_x - \varepsilon_y) \tag{5}$$

From Eqs.4 and 5, it follows that, the shear stiffness, G_{12}, can be determined from Eq.6, where the prefix Δ signifies a small increment.

$$G_{12} = \Delta(\sigma_x/2) / \Delta(\varepsilon_x - \varepsilon_y) \tag{6}$$

Finally, using Eq.4, the shear strength, S_{12}, can be obtained from Eq.7, where $P_{ult.}$ is the ultimate applied load.

$$S_{12} = P_{ult.}/2A \tag{7}$$

Since the ratio $\Delta\tau_{12}/\Delta\gamma_{12}$ is defined in ASTM D 3518 as being the slope of the shear stress/strain curve within the linear portion of the curve, the present experimental findings show that the value of the modulus is independent of specimen size. However, the shear strength, as defined by the standard, depends strongly upon the specimen thickness. Moreover, the choice of stacking sequence is left to the user, which according to the present experimental evidence, will also have a strong influence on the measured shear strength. Note that, two of the specimen size/stacking sequence combinations, **Ab** and **Bb**, which were studied herein, conform to the ASTM D3518, and **Ab** conforms to SACMA SRM-7 recommendations for specimen configuration. The shear strength, as calculated from the average $\sigma_{ult.}$ values in table 3, is underestimated by 4.5% if 8-ply specimens are used instead of 16-ply. An even greater difference occurs when the results from 8-ply specimens are compared to the results from 32-ply specimens. Focusing on the shear strength, the following discussion will attempt to isolate the reasons for such inconsistencies in the strength values and the dependency on specimen size, starting with the validity of the relationship between the ultimate axial stress and the in-plane shear strength.

Rosen [16] who proposed the relationship between the applied stress and the ply shear stress, stated that the local shear stress, τ_{12}, in each layer is independent of material properties and equals to one half of the applied stress, σ_x. However, problems may arise due to the fact that the plane of maximum shear stress for the laminate is different than that of the individual plies, where failure will initiate. As shown in figure 19, the plane of maximum shear stress for the laminate is orientated at 45°, to the loading direction; but, for the individual plies, the plane of maximum shear lies at an angle

$45\pm\theta°$, where the angle $\theta°$ is defined in figure 19. The difference occurs as a result of an in-plane shear stress, τ'_{xy}, which exists at the ply level only and is required for ply compatibility. Therefore, while Rosen's statement is correct, and holds for both the laminate and the ply level, the usefulness of Eq.7 depends upon the uniformity of σ_x, both across the width and through the thickness of the specimen. Clearly, if the average applied laminate stress σ_x is equal to the ply stress, σ'_x, then the average laminate shear stress, τ_{12}, will be equal to the ply shear stress, τ'_{12}. However, if $\sigma'_x > \sigma_x$ then $\tau'_{12} > \tau_{12}$ and hence the measured shear strength, S_{12}, will be underestimated. The ply stress σ'_x can be larger than the average laminate stress σ_x, due to the inhomogeneity of the composite laminate. For example, in every laminated composite, there exists a resin rich region between plies of different orientation. This region will have a much lower stiffness than the rest of the ply and hence it will carry less stress. Errors can also occur if the ply stress field is considerably altered by the presence of thermal residual stresses, which being self-equilibriating, will not affect the overall laminate stress field.

Small inaccuracies can also occur as a result of significant fiber scissoring which is a function of the shear strain. As discussed previously, for a Poisson's ratio $\nu_{xy} = 0.7$ and an applied axial strain of 2% the fiber rotation is about 1°. Note that, the 2% axial strain corresponds approximately to the point where the first peak in axial stress occurs, as shown in figure 3. Clearly there must be a strain limit beyond which Eq.7 cannot be applied due to large geometric changes. If the fibers are oriented at an angle other than 45°, then $\tau_{45°} = \sigma_{ult.}/2$ but $\tau_{12} \neq \sigma_{ult.}/2$. However, there is an even more important factor that can render Eq.7 invalid, which is the onset of damage. Once transply cracks develop, large local stresses will result and the local ply stresses can no longer be related to the applied stress in a simple manner. Moreover, the axial stiffness of the laminate will decrease at a rate which will depend upon the type of damage and the rate of damage accumulation. If all plies in the laminate crack simultaneously and damage evolves uniformly in all plies then the result will be a steady reduction in axial laminate stiffness, which may also cause fiber scissoring. If, on the other hand, damage occurs in one group of plies first, the laminate stiffness will be reduced; but, in addition, the laminate will become unbalanced and will be subjected to shear extension coupling. In either case the relationship between the normal and shear strengths, given by Eq.7, will be invalid.

In summary, the validity of Eq.7, for a balanced and symmetric ±45° laminate, depends upon;

(1) the uniformity of the ply stress, σ'_x, and its relationship to the laminate stress, σ_x,
(2) the fiber orientation, which for an elastic body is directly related to the shear strain, and
(3) the state of damage within the laminate.

So far, in the discussion of the validity of Eq.7, we have assumed a state of plane stress which is, of course, only valid for relatively thin and wide laminates. However, for practical reasons, typical laboratory size specimens are relatively narrow, and the three dimensional edge stresses may affect their mechanical behavior.

When the ±45° tensile test was proposed by Petit [17], he pointed out that, since there was no applied shear stress at the specimen edges, the ply stress τ_{xy}, must fall to zero at the free edge. He added that the non-uniformity of the in-plane shear stress can be neglected if a relatively large specimen width is used. Terry [13] attributed the differences in the stress/strain behavior between 16-ply laminates with stacking sequence $(+45°_4/-45°_4)_s$ and $(\pm 45°)_{4s}$ to edge effects. In addition to the non-uniformity of the in-plane shear stress, τ_{xy}, across the specimen width, it has been shown by Murthy and Chamis [12], through a 3-dimensional finite element analysis, that σ_x is also non-uniform across the specimen width. At the free edges the value of σ_x was approximately 80% of the far field stress. They have also shown that the relative magnitudes of the interlaminar stresses, τ_{xz} and σ_z depend upon the w/t ratio. For w/t = 4, $\tau_{xz} = \sigma_z$, and for w/t greater than 4, τ_{xz} remains constant, while σ_z diminishes, until it is practically zero for specimens with w/t=30.

Clearly, we can conclude that there is a complex distribution of three dimensional stresses at the free edge, the magnitude and nature of which depend upon the w/t ratio, and the stacking sequence. The question is; how important is the effect of the 3-dimensional stress state at the edge of the specimen? If indeed, there exists a significant edge effect, this should be reflected in the stress/strain response of the ±45° laminates with different w/t ratios. The results shown in figure 7, for the epoxy matrix specimens, show conclusively that there is no significant difference in the stress/strain response of such laminates. The **Aa** and **Ab** specimens which have a w/t ratio of 12.5 and 25.0 respectively, exhibit a very similar stress/strain response. Likewise, **Ca** and **Cc** specimens which have a w/t ratio of 4.2 an 12.5 respectively, exhibit a very similar stress/strain response up to 8.7% strain. In fact, the premature failure in the **Ca** specimen was associated with high gripping constraints due to the very narrow specimen area held in the jaws. During these tests, 2 out of 5 specimens failed within the jaw area and the other 3 failed very close to the jaws. A similar conclusion can be drawn from figure 8 for the PEEK matrix specimens. Moreover, it is shown in figure 3 that even though all four specimens **Aa**, **Bb**, **Cc**, and **Dd** share the same w/t ratio (=12.5), each one exhibits a different stress/strain response depending upon the number of plies. Specimens having the largest number of plies exhibited the greatest strength and strain to failure.

In addition to the validity of Eq.7, and the stress non-uniformity, it is important to define and determine correctly the value of the ultimate stress, in order to obtain a useful value of the shear strength. Ideally, $\sigma_{ult.}$ in Eq.7 should be the value of the applied axial stress at which in-plane shear failure occurs. A difficulty arises in that, failure initiates in the surface plies as a result of stresses normal to the fibers, and is followed by failure of the mid-plane plies. Thus, a $(\pm45°\pm45°)_{ns}$ laminate, for all values of n, contains four weak plies; two in either surface and a block of two plies in the mid-plane. The weakness in the surface plies results from the lack of support in the transverse-to-the-fibers direction which is supplied by neighboring plies in any other location within the laminate. As a result, shear strain, γ_{xy}, develops at the surface and close to the free edges, as indicated in figure 16, which leads to large normal strains, ε'_2. Therefore, a true shear strength as a material property cannot be determined unless the effect of the premature ply failures on the surface and in the mid-plane plies of the laminate are eliminated.

While the failure of the surface plies can be attributed to the non uniformity of the strain near the edge, this type of edge effect is not a function of w/t. Therefore, a given group of ±45° laminates, such as $(\pm45°)_{ns}$, will suffer similar free edge damage in their surface plies, irrespective of n and the w/t ratio. However, whether the damage is significant to the stress/strain response of the laminate, will depend upon n. The larger the total number of plies in a laminate, the smaller the effect of the pair of damaged surface plies will be.

The effect of premature failure of the blocked plies is also compatible with the 28.4% strength difference reported by Terry [13]. Clearly, the $(+45°_4/-45°_4)_s$ laminate is much more susceptible to premature failure both of the surface plies and the mid-plane plies when compared with the $(\pm45°)_{4s}$ laminates. Note that, the $(+45°_4/-45°_4)_s$ laminate, represents the case where all three blocks of plies (one on either surface and one in the middle) are susceptible to premature failure, (a) due to their position within the laminate and (b) due to the block thickness which results in reduced ply constraint and therefore lower *in situ* ply strength [15].

Based on the experimental findings, it is recommended that a stacking sequence with distributed plies be specified in the test standards with a minimum number of plies equal to 24. A minimum number of plies is required to eliminate the influence of specimen thickness on the shear strength measurement. In addition to the minimum specimen thickness it should be specified that the shear strength calculation be based on a representative applied stress value, such as the first peak stress and not the ultimate stress. While the above recommendations are valid for material systems with brittle matrices and/or strong fiber/matrix interfacial bond, materials with tough or compliant matrices like the AS4/PEEK system cannot be treated in a similar manner. For these material systems, there is no obvious point on the stress/strain curve that can be used in the definition of an in-plane shear strength. Therefore, the only remaining alternative is to use an off-set construction, based on some design parameter, such as the 0.4% strain. Since the stress/strain response of PEEK matrix specimens was also sensitive to the specimen thickness, it is recommended that a minimum number of plies (24 plies) would also be a requirement.

Note that, the first damage observed in the PEEK matrix specimens was thought to be fiber debonding which is a result of shearing. In this case it could be argued that the shear strength for such materials, with a relatively weak fiber/matrix interfacial bonding, is reached when the interface fails. Such strengths can still be determined from the stress/strain response of two or more specimen thicknesses, as the point where one deviates from the other. Such a technique could be used to characterize both the shear stress/strain response of a given material system, and the fiber/matrix interfacial strength. However, the appropriate specimen sizes have to be defined in order to eliminate possible interference with damage propagating in the free surface plies. It is believed that this subject merits further investigation.

Since normal tensile strains transverse to the fibers are responsible for first ply failures in the surface plies, it is anticipated that such premature failures could be suppressed if a compressive instead of a tensile stress is applied. Thin specimens (8 plies for example) tested in compression could produce reasonable values for the shear strength, and, at the same time, offer substantial material savings. Such a ±45° compressive test could be particularly attractive in cases were material shortages outweigh the additional difficulties involved with a compressive test.

5. CONCLUSIONS

5.1 First Ply Failure/Epoxy System

First ply failure in the epoxy matrix specimens with distributed plies occurred in the form of transply cracks, in the surface plies, as a result of primarily normal and <u>not</u> shear fracture. Likewise, second ply failure which occurred in the middle block of two -45° plies was also the result of primarily normal rather than shear fracture. It was shown that the first-ply-failure stress, for a given generic ±45° lay-up, depends primarily on the specimen thickness and not on the specimen width. The larger the number of distributed plies the greater the first-ply-failure stress. In the case of specimens with blocked plies, the first-ply-failure stress decreased as the thickness of the blocked plies was increased.

5.2 First Ply Failure/PEEK System

In PEEK matrix specimens, damage was thought to initiate in the form of fiber debonding. Such damage, which was visible in the enhanced X-ray radiographs but could not be verified by optical microscopy, occurred at a relatively low applied stress, much lower than the first-ply-failure stress in the corresponding epoxy specimens. Evidence of transply cracks was first observed to occur in the surface plies followed by cracks in the middle plies. These transply cracks occurred at stresses which were still lower than the first-ply-failure stress in the corresponding epoxy matrix specimens. As in the case of the epoxy matrix specimens, the stress value at which transply cracking initiated was primarily a function of the specimen thickness and not the specimen width.

5.3 Stress/strain Response/Epoxy System

As a result of the sensitivity of the first-ply-failure stress to the specimen thickness, a comparison of the stress/strain response of different thickness specimens could be used as an indicator of damage initiation. The deviation of a given stress/strain curve from the curve of a second geometrically scaled specimen corresponded to the point at which damage could be detected by non-destructive damage examination.

In general, the stress/strain response of the epoxy matrix specimen depended primarily upon two factors; the generic lay-up type, and the laminate thickness. Specimens with blocked plies exhibited a brittle-like stress/strain response and their ultimate strength and strain at failure increased with decreasing blocked ply thickness. Specimens with distributed plies, however, exhibited a ductile like stress/strain response and the ultimate strength and strain at failure increased with increasing specimen thickness. The specimen width had no significant influence on the stress/strain response.

The stress/strain response of the epoxy specimens with distributed plies was characterized by a region in which the gradient of the stress/strain curves became zero and then negative before it became positive again at slightly higher strains. This behavior was more obvious in specimens with 24 and 32 plies.

5.4 Stress/strain Response/PEEK System

The stress/strain response of 8-ply PEEK matrix specimens was found to be very sensitive to the distance of a given specimen from the panel edges. In comparison with the corresponding epoxy matrix specimens, the PEEK matrix specimens exhibited a more ductile stress/strain response, having a lower initial longitudinal stiffness and failing at higher strains. In the case of the baseline (8-ply) specimens, the difference in the strain at failure between the epoxy and the PEEK matrix specimens was approximately 15%. Another basic difference in the stress/strain response of the two material systems took place at a strain of approximately 2%, where the gradient of the stress/strain response of the epoxy matrix specimens became zero and then negative. This behavior was not observed in the PEEK specimens.

5.5 Strength and Strain at Failure/Epoxy Matrix

Both the ultimate strength and the strain at failure were a function of the generic lay-up (blocked or distributed plies) and the specimen thickness. Specimens with distributed plies showed an increase in ultimate strength and strain at failure with increasing specimen thickness. In specimens with blocked plies the ultimate strength and the strain at failure were reduced as the ply increased.

5.6 Strength and Strain at Failure/PEEK Matrix

Similar behavior to that of the epoxy matrix specimens was also observed in the PEEK matrix specimens. The ultimate strength and strain at failure increased with increasing specimen thickness. The ultimate strength and strain at failure of PEEK matrix specimens was always greater than the ultimate strength and strain at failure of corresponding epoxy matrix specimens.

5.7 The ±45° Standard Shear Test

This work has highlighted the need to redefine the specimen configuration in the ASTM D 3518 test if a representative value of the shear strength, as a material property, is to be determined. It is recommended that the minimum specimen thickness be changed from 0.5 mm to 3.0 mm (or from 4 to 24 plies) and that the stacking sequence be clearly specified; $(\pm45°/\pm45°)_{nS}$, where the value of n is at least 3.

5.8 Failure Criteria

This work has pointed out the difficulties involved in measuring the in-plane shear properties of a composite laminate. Consequently great care and good judgement must be exercised when shear strength or strain at failure values are obtained and subsequently used to predict the failure of a composite laminate.

6. SUGGESTED FUTURE WORK

Since first ply failure was found to occur in the surface plies, further studies are needed to develop a relationship between the surface ply displacements, the ply thickness, the fiber orientation and the material elastic properties, so that damage tolerant composite laminates, with surface angle plies can be designed.

The present study of the failure processes in angle ply laminates, under tensile loading, should be carried over to more practical types of laminates and loading conditions such as, for example, compressive loading. In particular, the ±45° tensile test should be reassessed under compressive loading, which may provide a more efficient and accurate method for determining the in-plane shear strength of advanced composite materials.

The mechanisms of first and second ply failures, which were observed in the present studies, should be incorporated into appropriate failure theories to predict the ultimate strength of angle ply laminated composites.

The possibility of employing the techniques for correlating first ply failure to the stress/strain responses of scaled specimens, for sensitive non-destructive damage evaluation, should be pursued further.

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[17] Petit P.H., "A Simplified Method of Determining the In-plane Shear Stress/Strain Response of Unidirectional Composites", Composite Materials: Testing and Design, ASTM STP 460, 1969, pp. 83-93.

[18] Pipes R.B and Daniel I.M., "Moire' Analysis of the Interlaminar Shear Edge Effect in Laminated Composites", Journal of Composite Materials, Vol.5, 1971, pp. 255-259.

Table 1. Test matrix for AS4/3502 indicating the specimen dimensions and the type of lay-ups tested in each case. Note that specimens **Aa, Bb, Cc,** and **Dd** are scaled in three dimensions.

Nominal Thickness **mm**	Nominal Width x Nominal Length **mm** x **mm**			
	12.7 x 127 (Size **a**)	**25.4 x 254** (Size **b**)	**38.1 x 381** (Size **c**)	**50.8 x 508** (Size **d**)
1.0 (8-plies) Lay-up **A**	**Baseline**	/ Distributed	/ /	/ /
2.0 (16-plies) Lay-up **B**	/ Distributed	Blocked[1] Distributed	/ /	/ /
3.0 (24-plies) Lay-up **C**	/ Distributed	/ /	Blocked[1] Distributed	/ /
4.0 (32-plies) Lay-up **D**	/ /	Blocked Distributed	/ /	Blocked[1] Distributed

[1] These tests are reported in reference [6].

Table 2. Test matrix for AS4/PEEK indicating the specimen dimensions and the type of lay-up tested in each case. Note that specimens **Aa**, and **Bb** are scaled in three dimensions.

Nominal Thickness **mm**	Nominal Width x Nominal Length **mm** x **mm**	
	12.7 x 127 (Size **a**)	**25.4 x 254** (Size **b**)
1.0 (8-plies) Lay-up **A**	**Baseline**	/ Distributed
2.0 (16-plies) Lay-up **B**	/ Distributed	/ Distributed

Table 3: Average ultimate stress for sublaminate scaled AS4/3502 specimens.

Size (mm)	Average Ultimate Stress - MPa (Coefficient of Variation)			
	A-(±45°/±45°)$_s$	B-(±45°/±45°)$_{2s}$	C-(±45°/±45°)$_{3s}$	D-(±45°/±45°)$_{4s}$
a (12.7x127)	159.34 (3.05)	167.54 (1.52)	197.40 (2.03)[1]	N/A
b (25.4x254)	159.27 (2.08)	166.78 (2.65)	N/A	N/A
c (38.1x381)	N/A	N/A	227.32 (3.78)	N/A
d (50.8x508)	N/A	N/A	N/A	248.62 (2.00)

[1] As a result of the low w/t = 4.2, exceptionally high stress concentrations due to gripping constraints obstructed true gage failures from occurring. Only three out of five replicate test results for ultimate stress and strain were deemed acceptable.

Table 4: Average strain corresponding to ultimate stress for sublaminate scaled AS4/3502 specimens.

Size (mm)	Average Ultimate Strain - % (Coefficient of Variation)			
	A-(±45°/±45°)$_s$	B-(±45°/±45°)$_{2s}$	C-(±45°/±45°)$_{3s}$	D-(±45°/±45°)$_{4s}$
a (12.7x127)	1.61 (6.14) 1.86 (5.97)[1]	N/A	9.30 (9.14)	N/A
b (25.4x254)	1.78 (4.49)	1.82 (5.86) 1.97 (6.08)[2]	N/A	N/A
c (38.1x381)	N/A	N/A	10.66 (4.67)	N/A
d (50.8x508)	N/A	N/A	N/A	10.83 (4.39)

[1] The displacement in 8 out of 16 specimens was monitored with a custom built extensometer with a gage length of 38 mm

[2] The displacement in 4 out of 8 specimens was monitored with a custom built extensometer with a gage length of 76 mm.

Table 5: Average applied stress at the first turning point for sublaminate scaled specimens.

Size (mm)	Average Peak Stress - MPa (Coefficient of Variation)			
	A-(±45°/±45°)$_s$	B-(±45°/±45°)$_{2s}$	C-(±45°/±45°)$_{3s}$	D-(±45°/±45°)$_{4s}$
a (12.7x127)	N/A	N/A	165.33 (2.34)	N/A
b (25.4x254)	N/A	N/A	N/A	N/A
c (38.1x381)	N/A	N/A	170.37 (2.67)	N/A
d (50.8x508)	N/A	N/A	N/A	182.09 (0.26)

Table 6: Average strain corresponding to the first turning point for sublaminate scaled specimens.

Size (mm)	Average Strain - % (Coefficient of Variation)			
	A-(±45°/±45°)$_s$	B-(±45°/±45°)$_{2s}$	C-(±45°/±45°)$_{3s}$	D-(±45°/±45°)$_{4s}$
a (12.7x127)	N/A	N/A	2.19 (12.26)	N/A
b (25.4x254)	N/A	N/A	N/A	N/A
c (38.1x381)	N/A	N/A	2.37 (4.09)	N/A
d (50.8x508)	N/A	N/A	N/A	2.43 (4.28) 2.46 (3.90)[1]

[1] Values from custom built extensometer with a gage length of 152 mm, used in conjunction with the MTS extensometer.

Table 7. Average ultimate stress for sublaminate scaled AS4/PEEK specimens.

Size (mm)	Aver.[1] Ultimate Stress - MPa (Coefficient of Variation)	
	A-(±45°/±45°)$_s$	B-(±45°/±45°)$_{2s}$
a (12.7x127)	250.26 (13.84) 279.80 (4.23)[2]	362.19 (2.72) 365.73 (0.09)[4]
b (25.4x254)	312.23 (17.82) 289.53(16.71)[3]	378.24 (1.29)

[1] Average from at least 7 duplicate tests.
[2] Average of three specimens closest to the middle of the panel.
[3] Average of three specimens chosen from an equivalent panel position to that of the **Aa** specimens.
[4] Average value of ultimate stress excluding the specimen which came from the end of the panel.

Table 8. Average strain corresponding to ultimate stress for sublaminate scaled AS4/PEEK specimens.

Size (mm)	Average Ultimate Strain[1] - % (Coefficient of Variation)	
	A-(±45°/±45°)$_s$	B-(±45°/±45°)$_{2s}$
a (12.7x127)	15.95 (11.10)	19.97 (4.63)
b (25.4x254)	17.74 (15.45)	21.13 (2.69)

[1] These represent extrapolated values, since the extensometer range was only 15.0% strain. In the extrapolation the measured values of stress were used in conjunction with an assumed constant rate of strain.

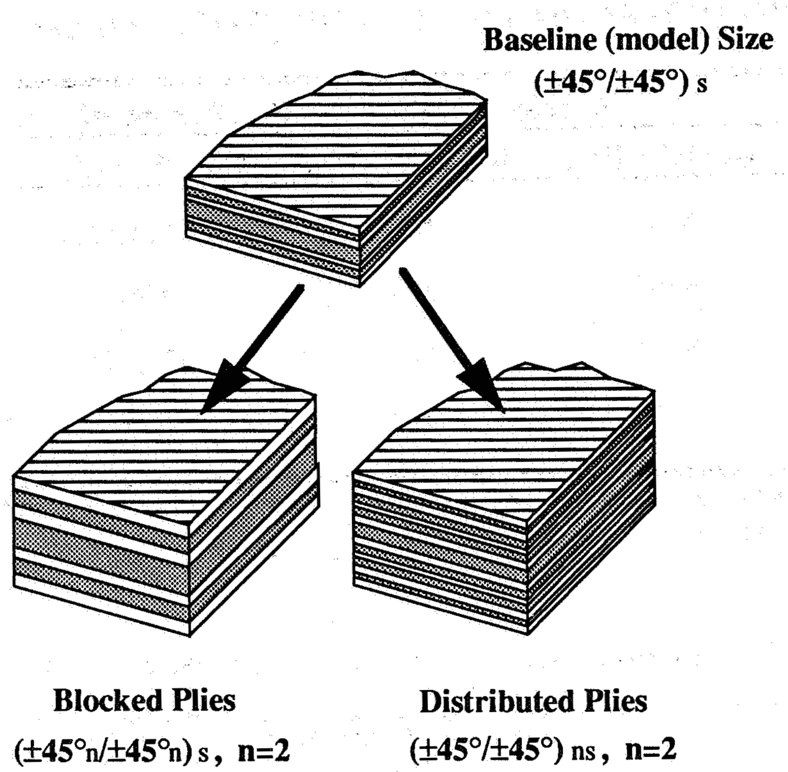


Fig.1 Schematic of thickness build-up procedures for composite laminates. Note that; while in both techniques the extensional stiffness remains constant with increasing laminate thickness, the bending stiffness will remain constant only in the case of ply level scaled laminates.

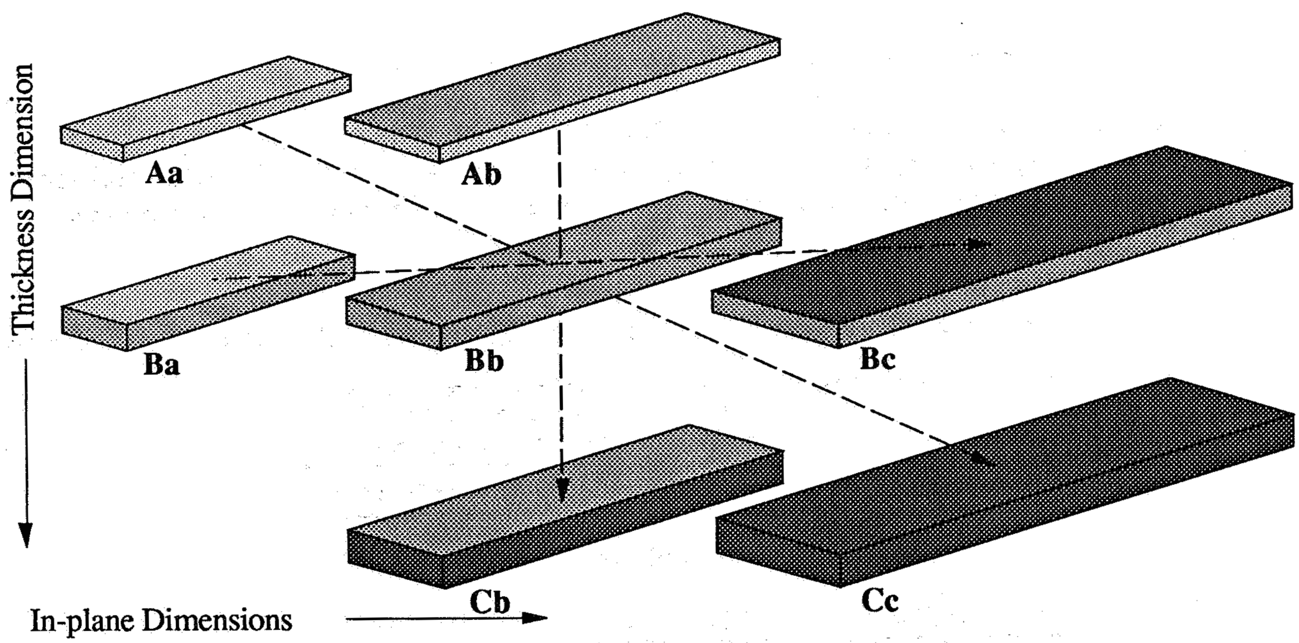


Fig.2 Schematic of dimensional scaling procedures for composite laminates. The thickness dimension is increased according to one of the two methods shown in figure 1. Note that; specimens which lie on the diagonal line (**Aa**, **Bb**, and **Cc**) represent three dimensional scaling.

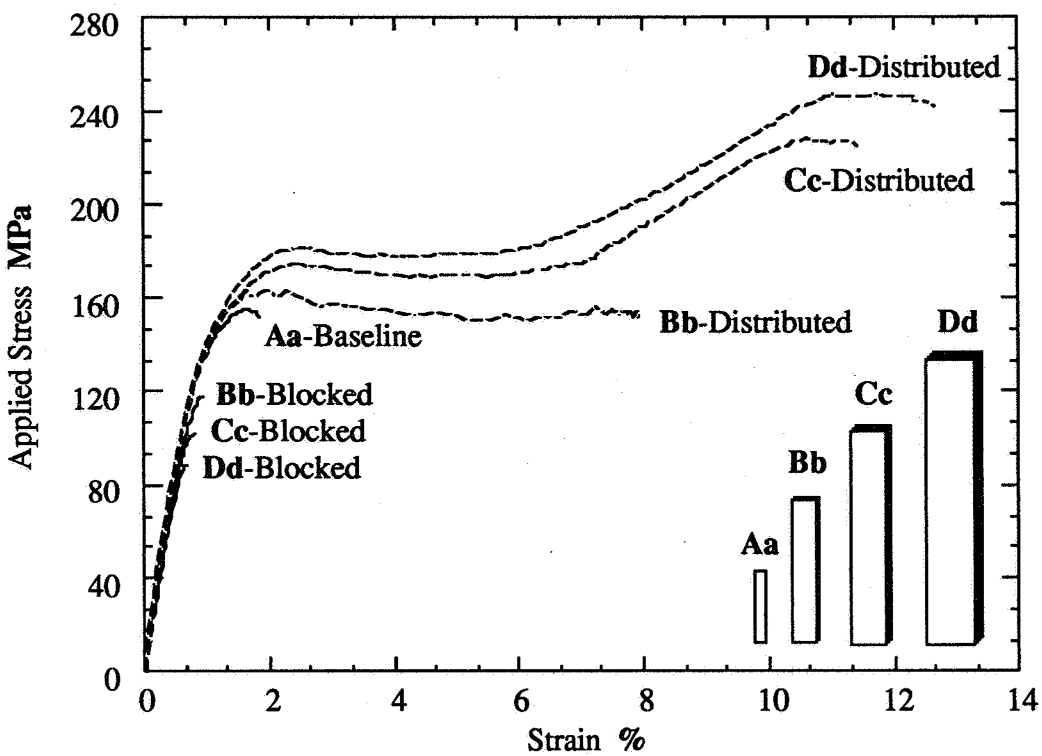


Fig.3 Stress/strain response of epoxy matrix specimens scaled in three dimensions. **Aa**, **Bb**, **Cc** and **Dd** are equal to; 12.7xn mm wide, 127xn mm long and 1xn mm thick where n= 1, 2, 3, and 4) respectively.

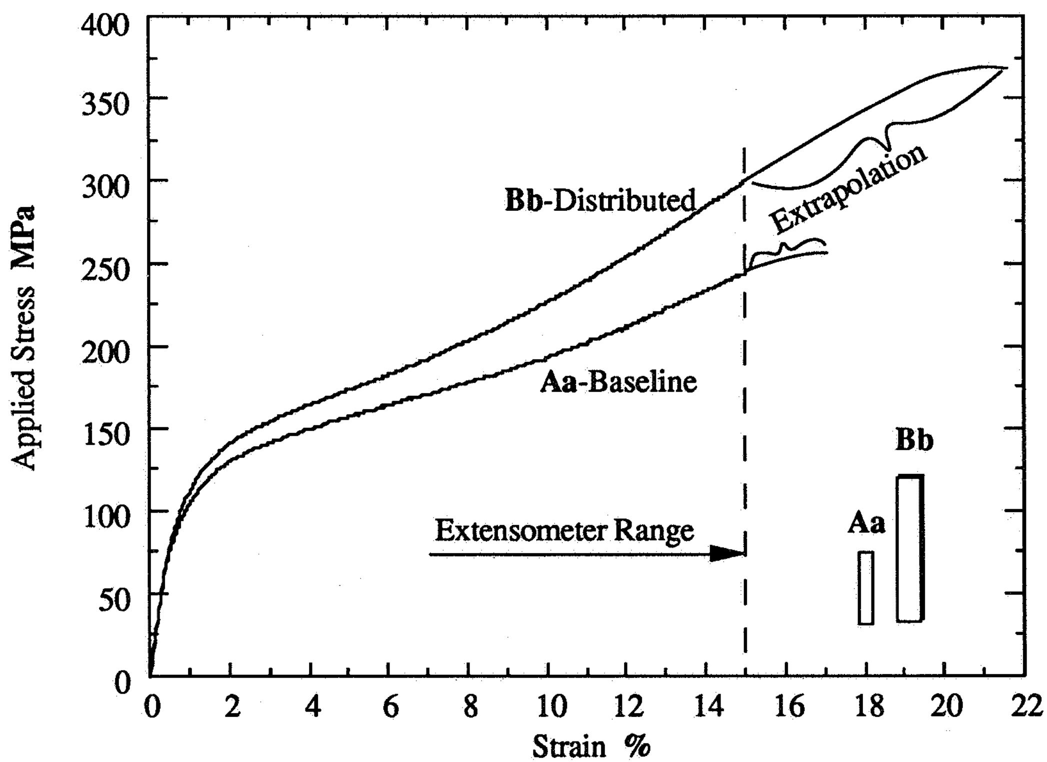


Fig.4 Stress/strain response of PEEK matrix specimens scaled in three dimensions. Note that the extensometer range of 15% was exceeded. Beyond this point, both curves were extrapolated to the maximum measured stress.

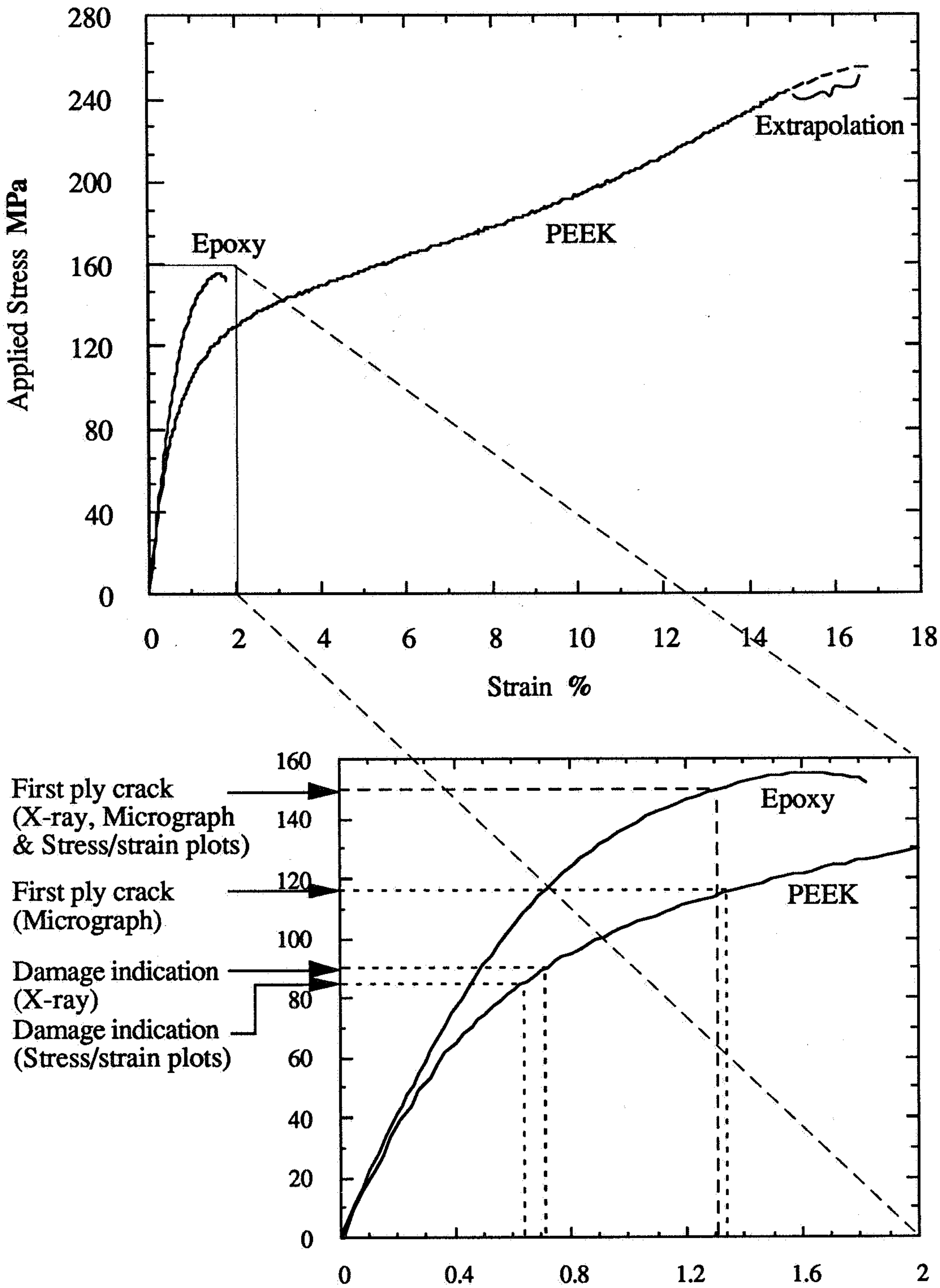


Fig.5 Comparison of the stress/strain response of epoxy and PEEK matrix, baseline, specimens.

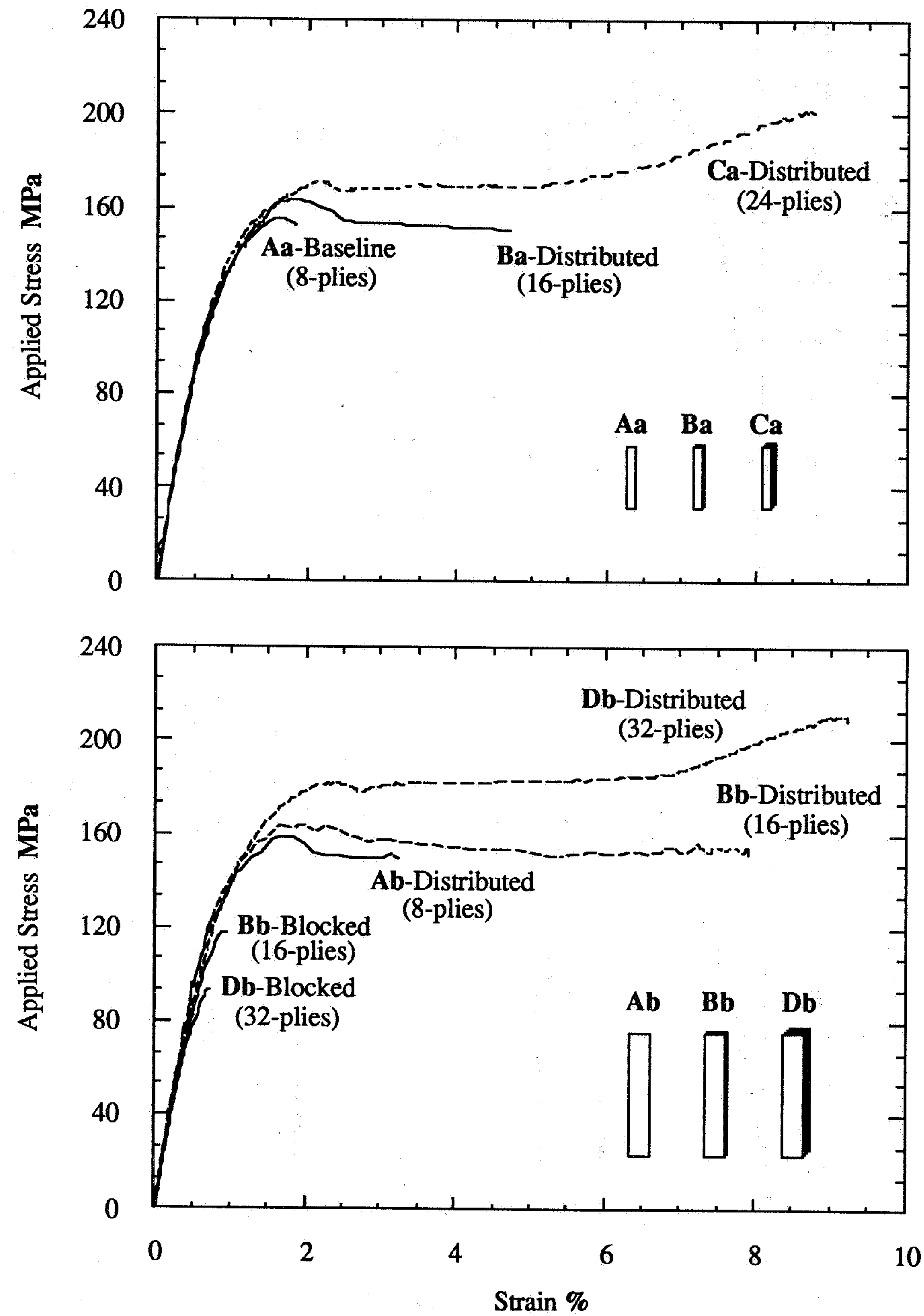


Fig.6 Stress/strain response of epoxy specimens scaled in one dimension. The in-plane dimensions are kept fixed: **(a)** 12.7 x 127 mm (top), and **(b)** 25.4 x 254 mm (bottom).

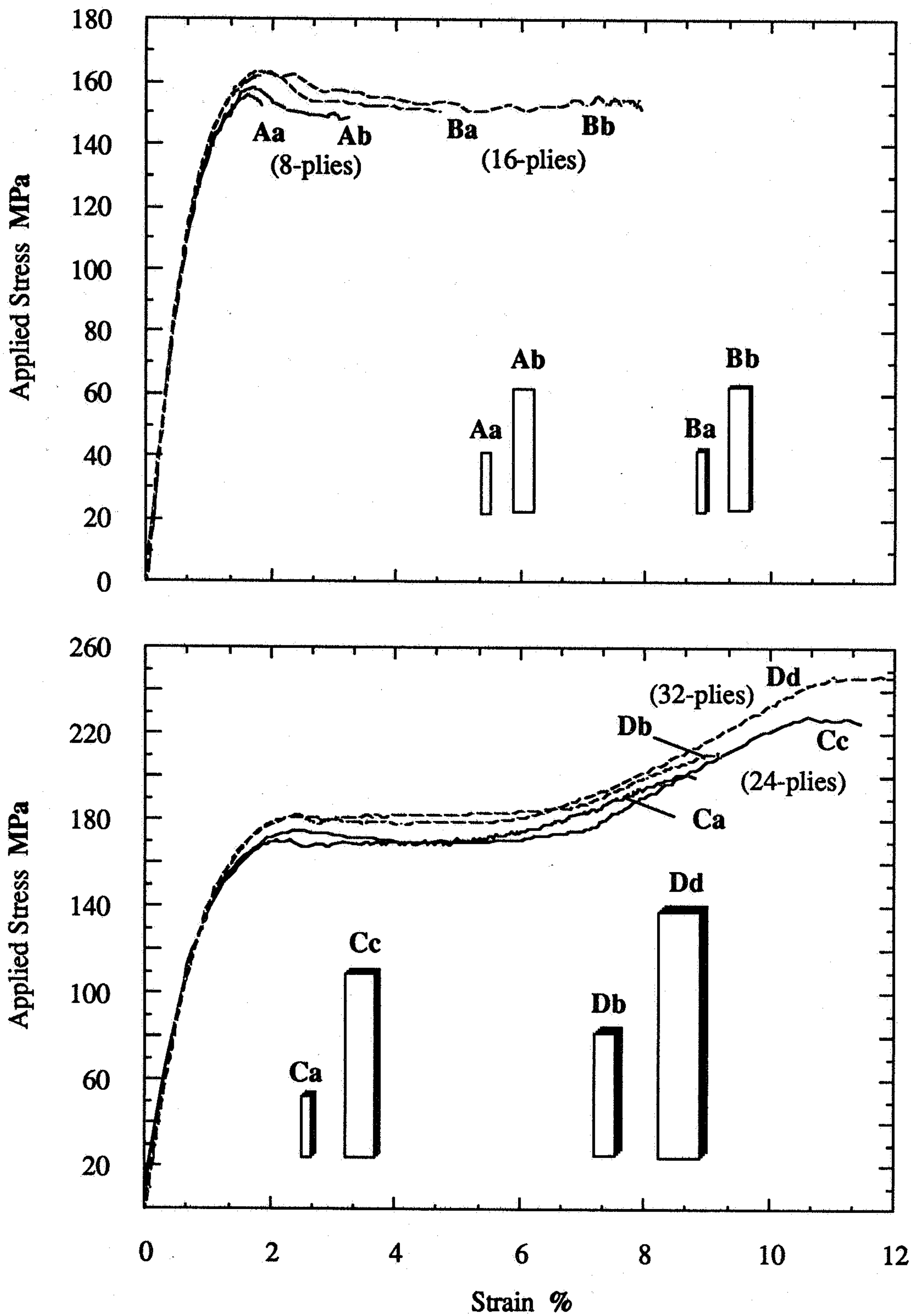


Fig.7 Stress/strain response of epoxy matrix specimens scaled in two dimensions. The thickness is kept fixed: **(a)** 1 or 2 mm (top) and **(b)** 3 or 4 mm (bottom).

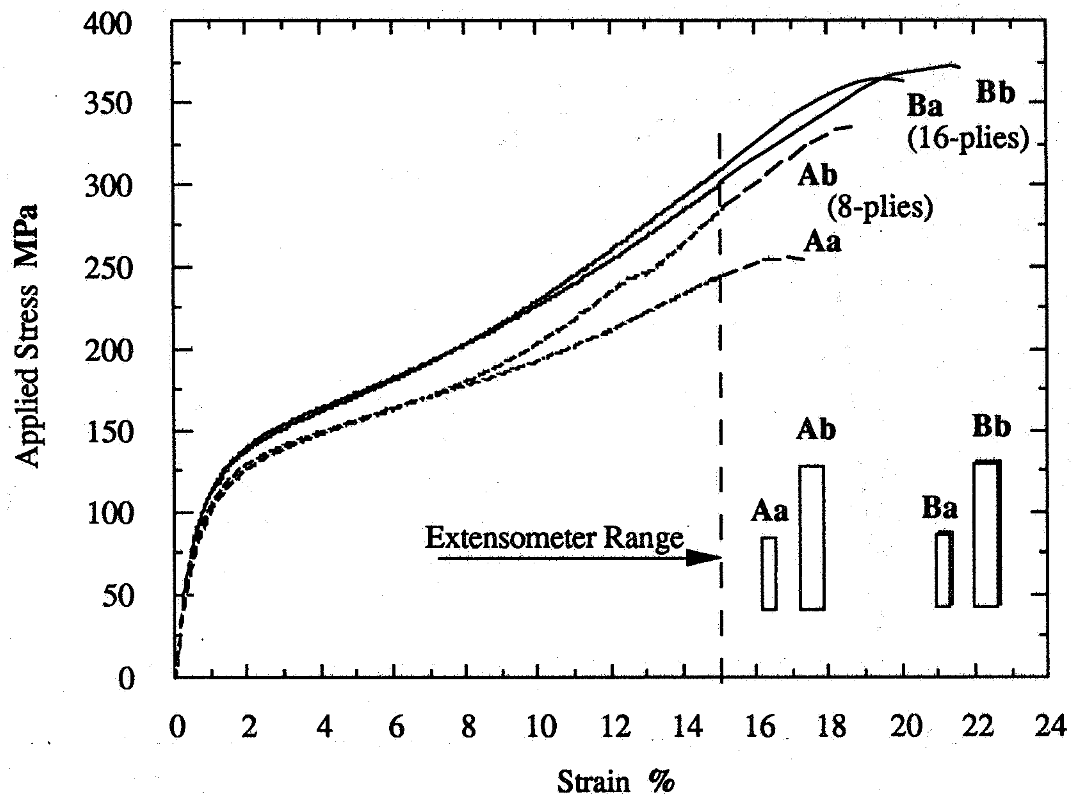


Fig.8 Stress/strain response of PEEK matrix specimens scaled in one and two dimensions. Note that the extensometer range of 15% was exceeded. Beyond this point all curves were extrapolated to the maximum measured stress.

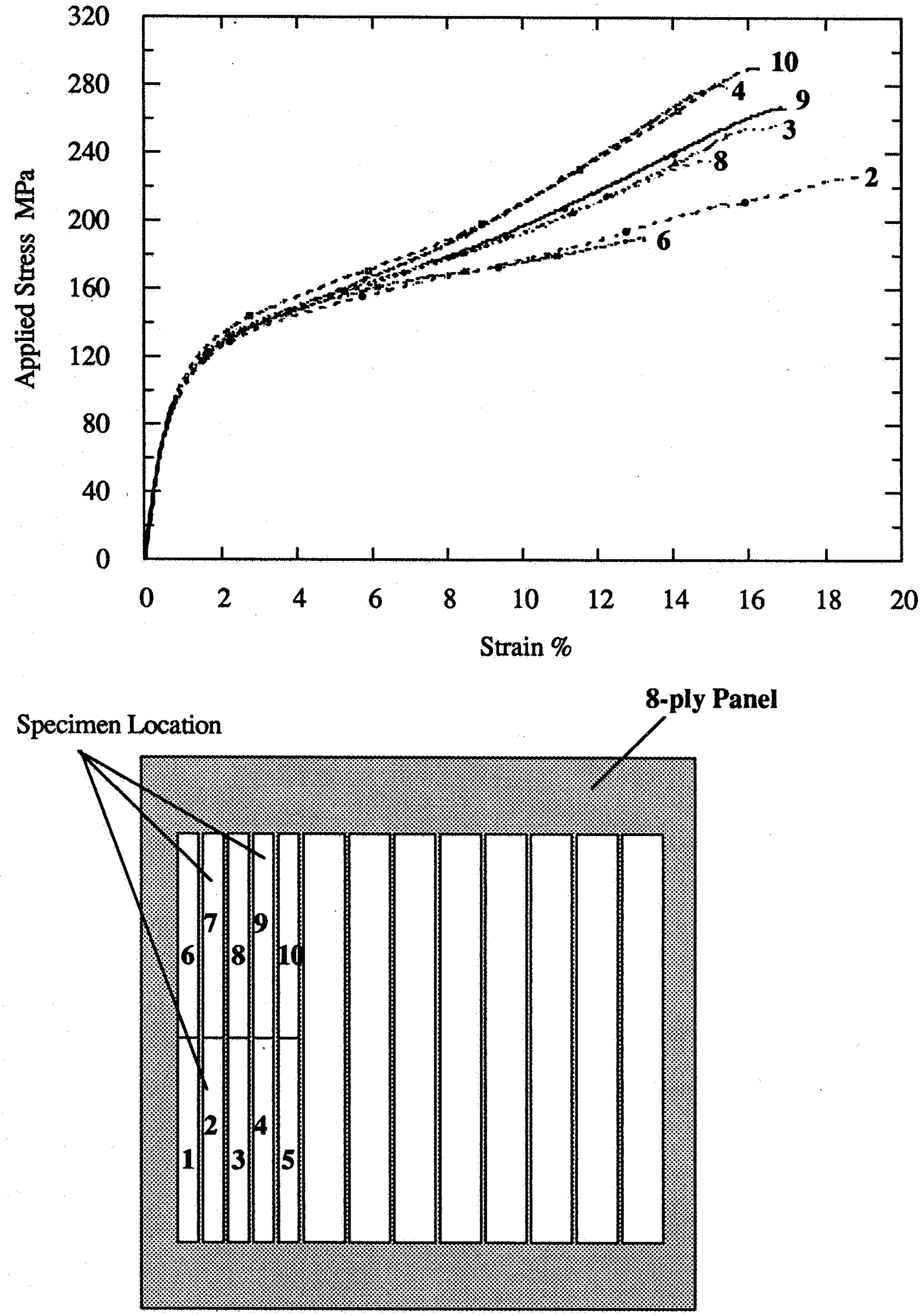


Fig.9 Stress/strain responses of 8-ply, size a, PEEK matrix specimens as a function of specimen position.

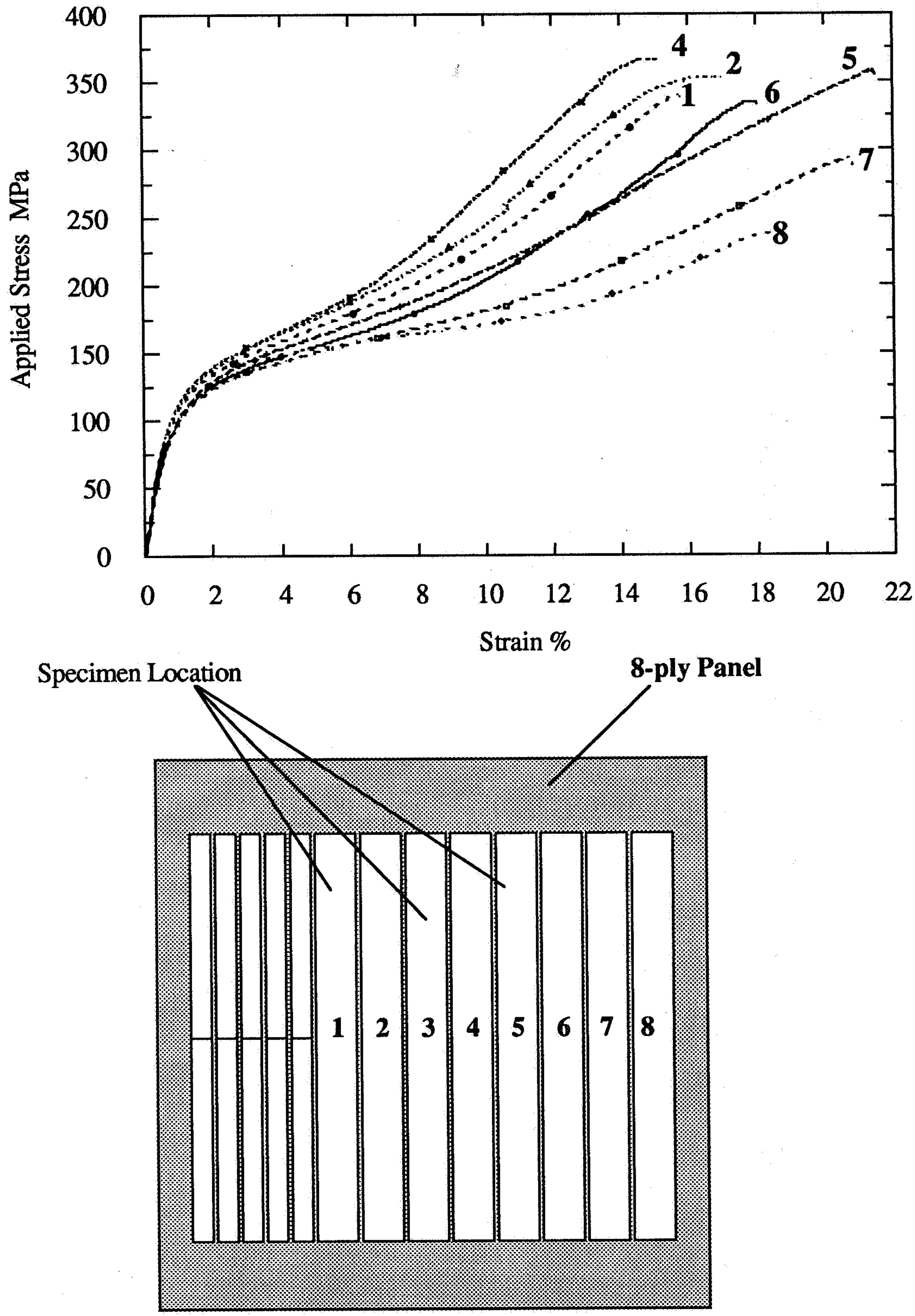


Fig.10 Stress/strain responses of 8-ply, size **b**, PEEK matrix specimens as a function of specimen position.

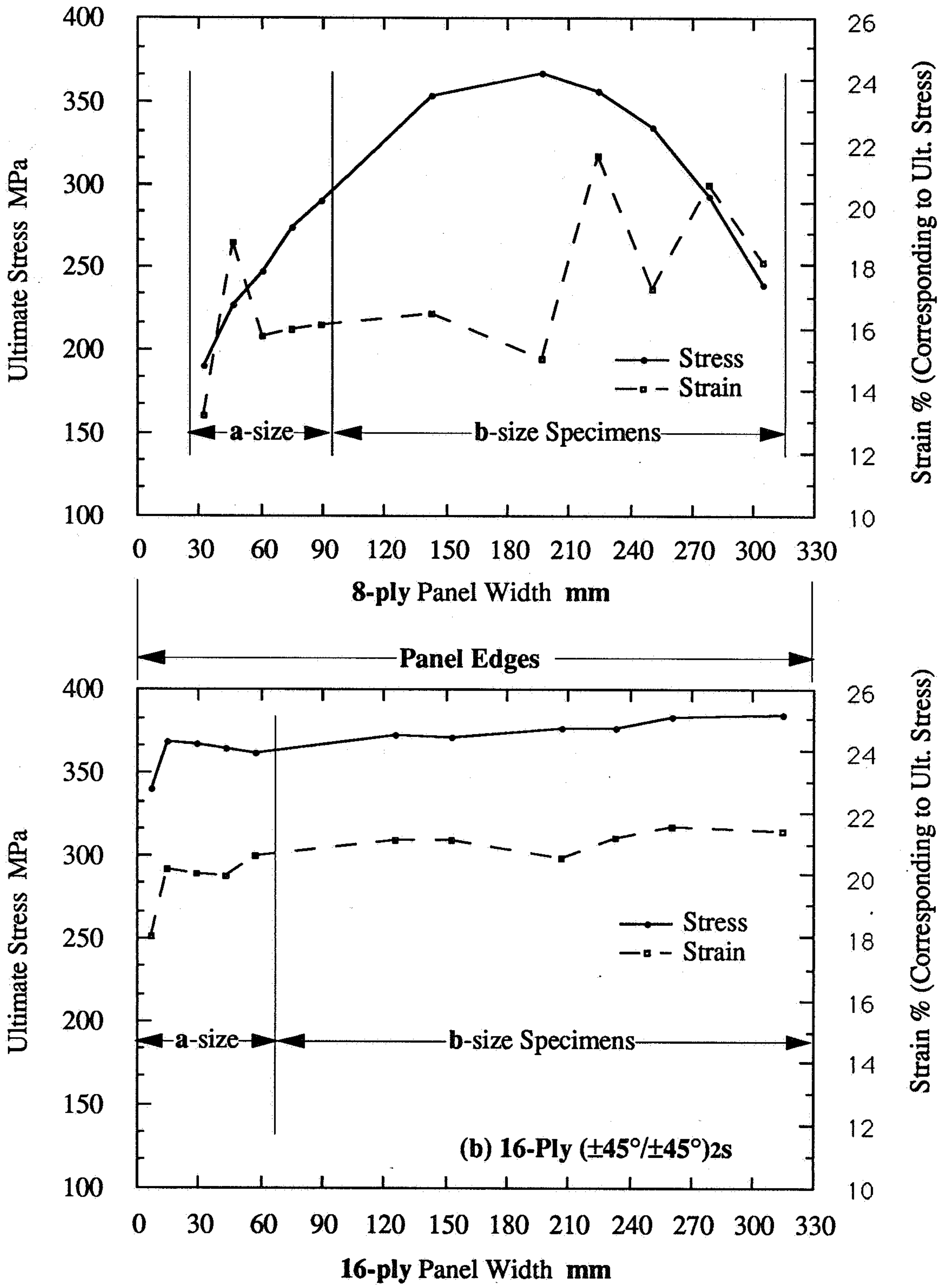
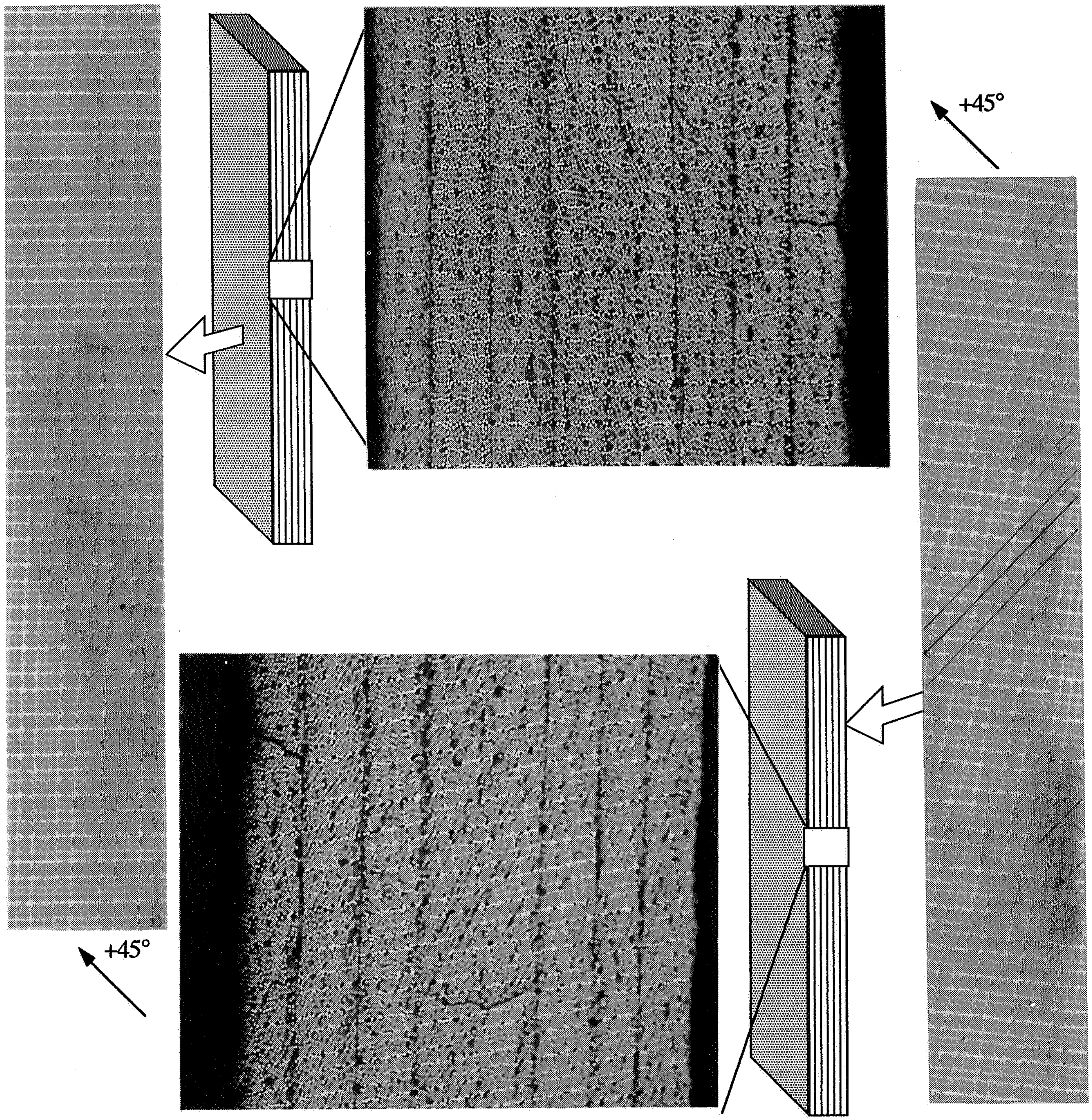


Fig.11 Ultimate stress and strain at failure versus specimen position, for 8 and 16-ply PEEK matrix specimens.

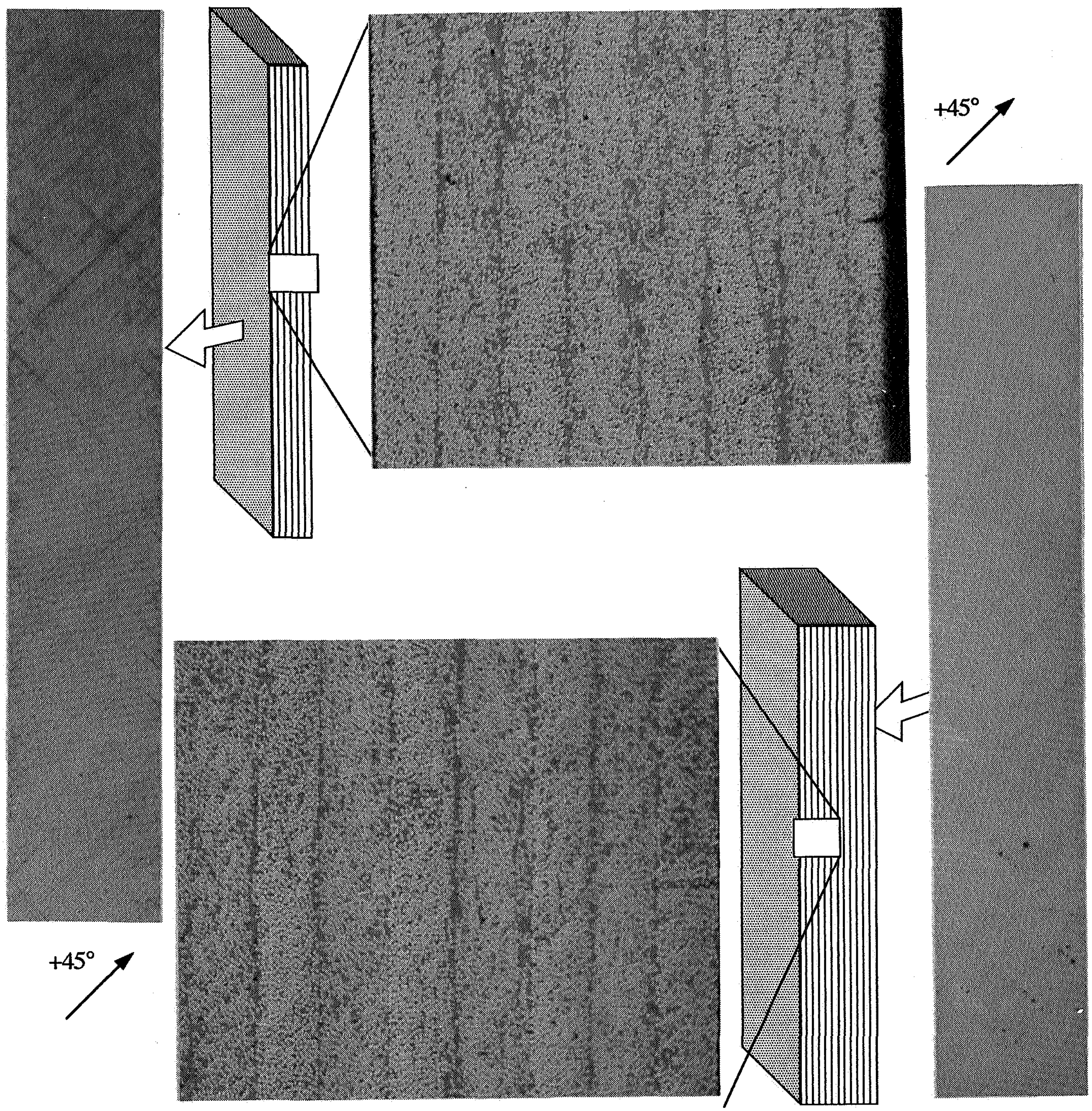
(a) 8-ply Epoxy Matrix - Loaded to **150 MPa**



(b) 8-ply Epoxy Matrix - Loaded to **158 MPa**

Fig.12 First and second ply failures for 8-ply epoxy matrix shown in **(a)** and **(b)**, respectively.

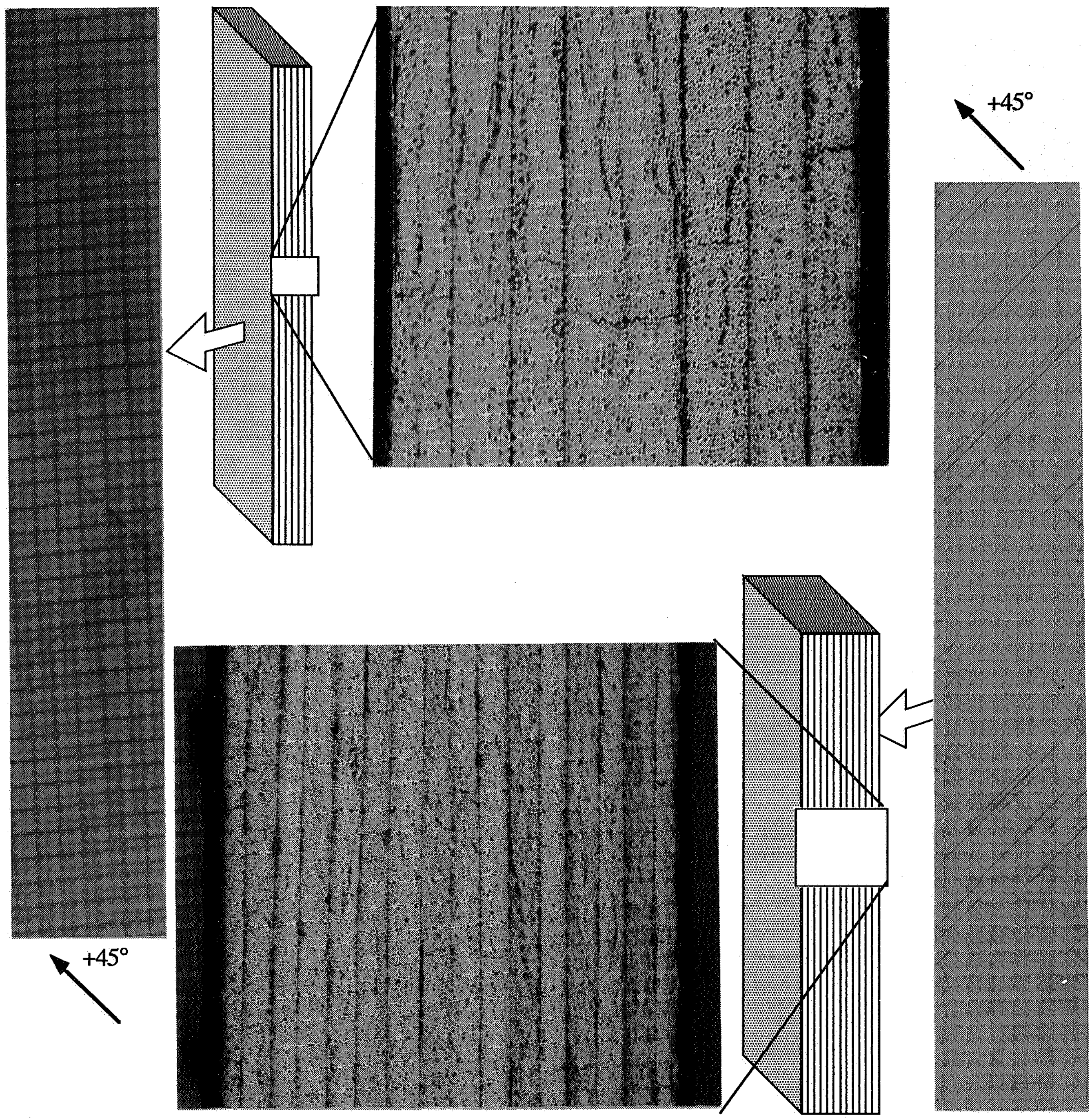
(a) 8-ply PEEK Matrix - Loaded to **115 MPa**



(b) 16-ply PEEK Matrix - Loaded to **131 MPa**

Fig.13 First ply cracking (visible by an optical microscope) for 8 and 16-ply PEEK matrix specimens, shown in **(a)** and **(b)** respectively. Note that, while transply cracks are only visible on the surface (+45°) plies, the X-ray radiographs indicate that there is damage in both +45° and -45° directions.

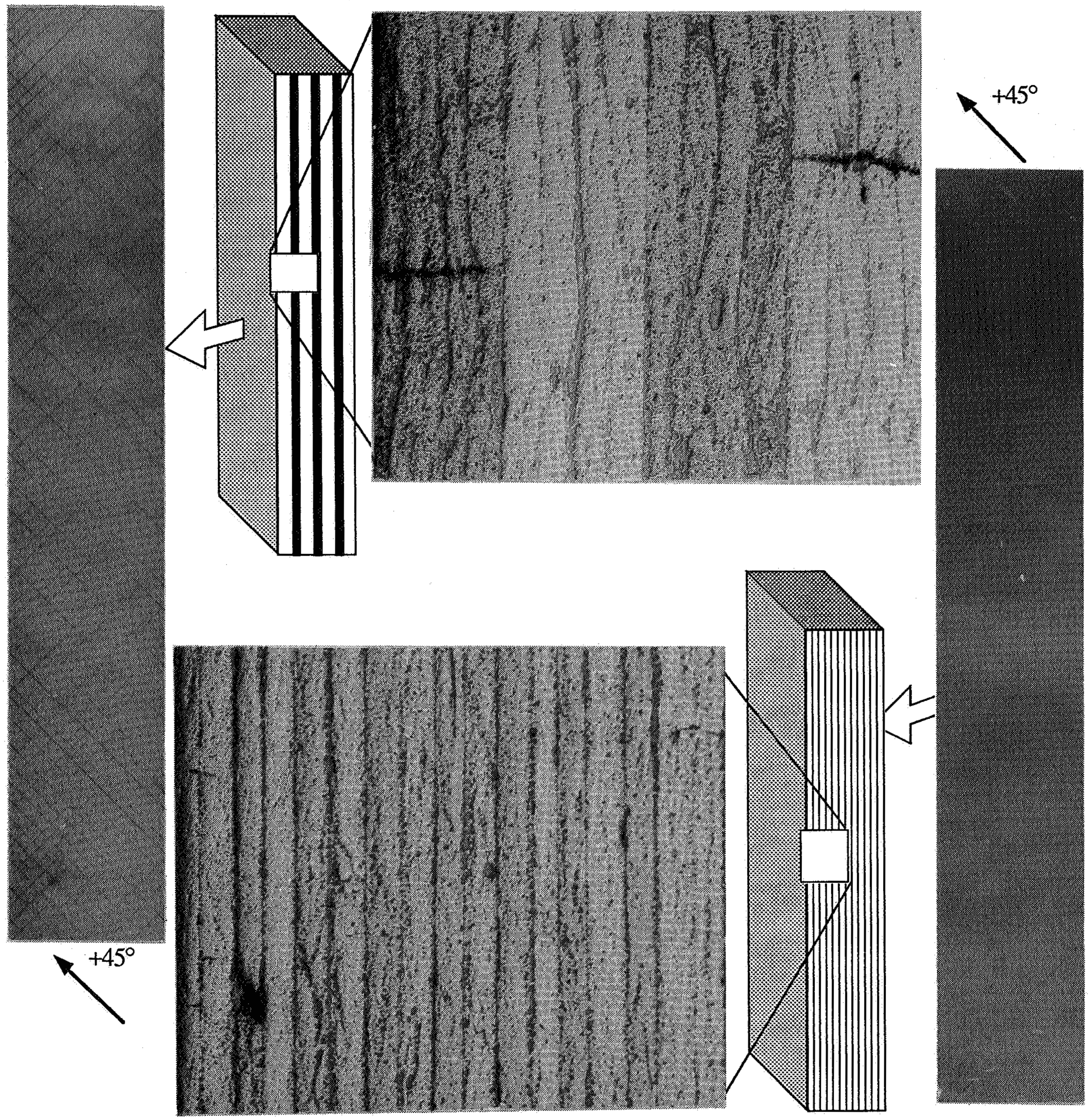
(a) 8-ply Epoxy Matrix - Loaded to **158 MPa**



(b) 16-ply Epoxy Matrix - Loaded to **164 MPa**

Fig.14 Developed damage states in 8 and 16-ply epoxy matrix specimens shown in **(a)** and **(b)** respectively. Note that, damage in the 8-ply specimen is very localized whereas in the 16-ply specimen it is well distributed.

(a) 32-ply Epoxy Matrix Specimen with Blocked plies - Loaded to **0 MPa**



(b) 32-ply Epoxy Matrix Specimen with Distributed Plies- Loaded to **180 MPa**

Fig.15 A comparison between the damage states in ply and sublaminate level scaled, 32-ply epoxy matrix specimens, shown in **(a)** and **(b)** respectively.

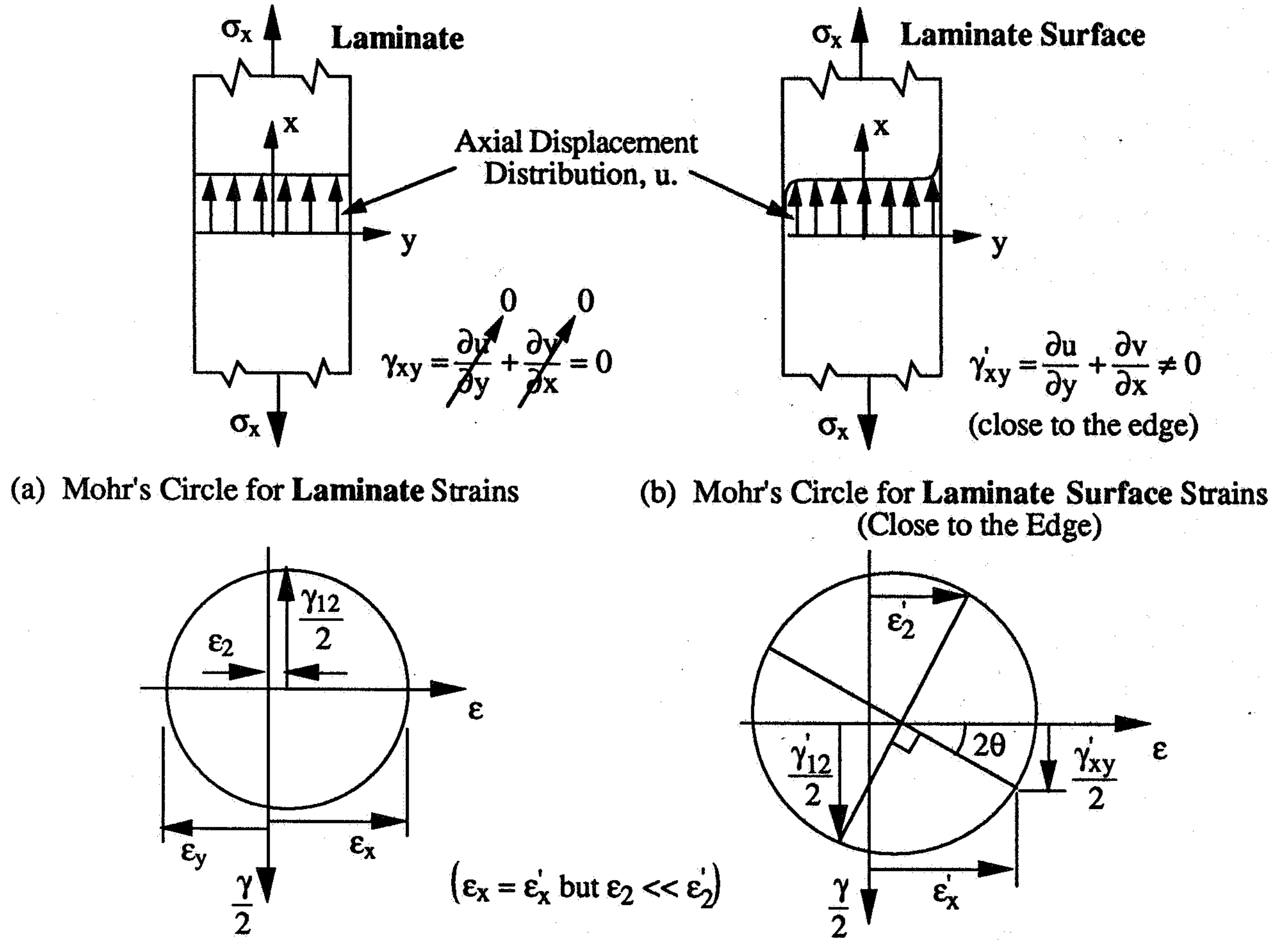
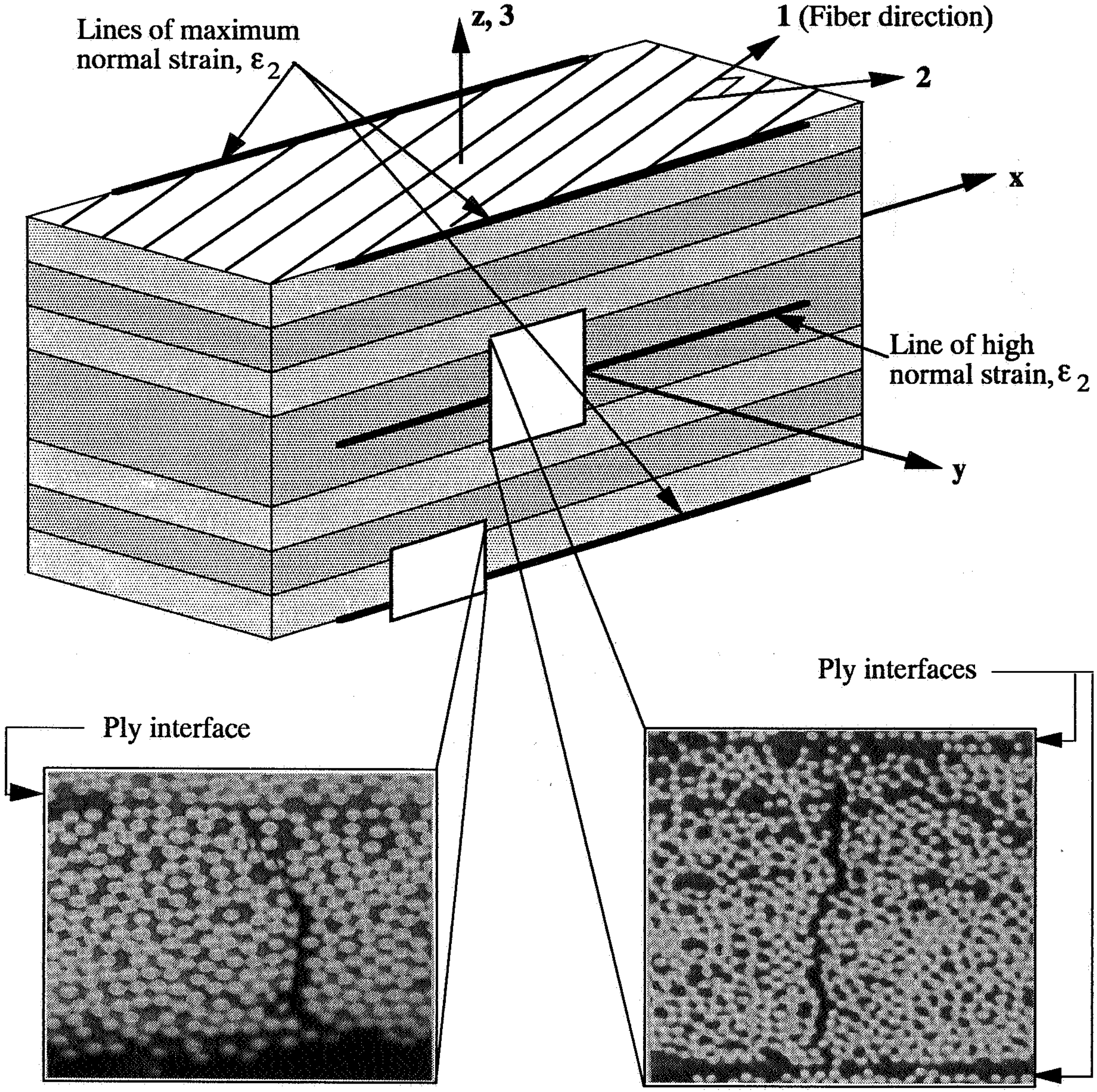


Fig.16 The effect of the surface shear strain γ_{xy}, at the free edges, after Pipes and Daniel [18]. Note that, at the corners where the laminate surface meets with the laminate free edges, ε'_2 is much greater than the strain ε_2 anywhere else in the laminate.



(a) First-ply failure **(b) Second ply failure**

Fig.17 Close-up view of first and second ply failures in epoxy matrix specimens. Laminate and ply axes: x-y-z are the laminate axes with x being along the loading direction. 1-2-3 are the ply axes with 1 along the fiber direction.

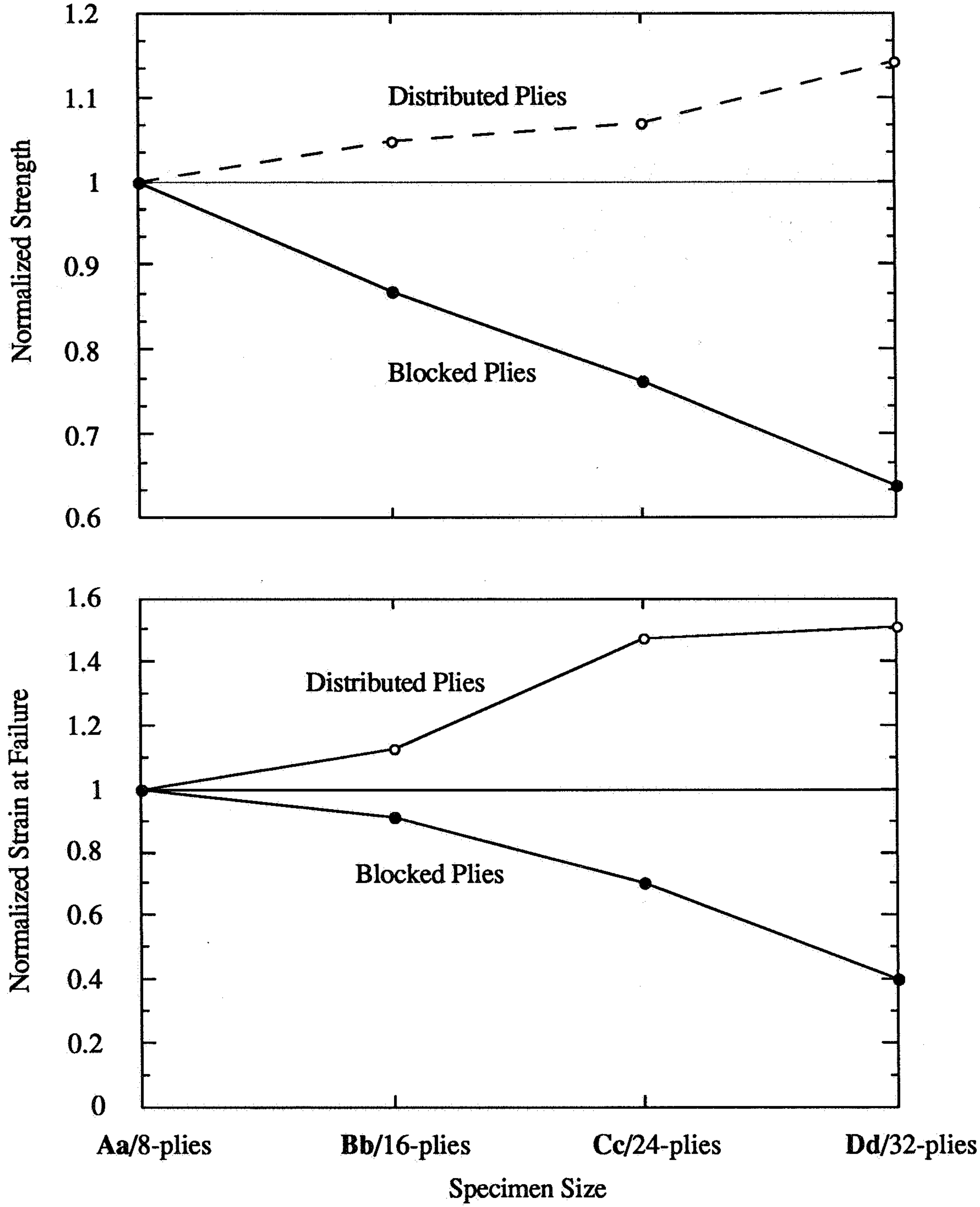
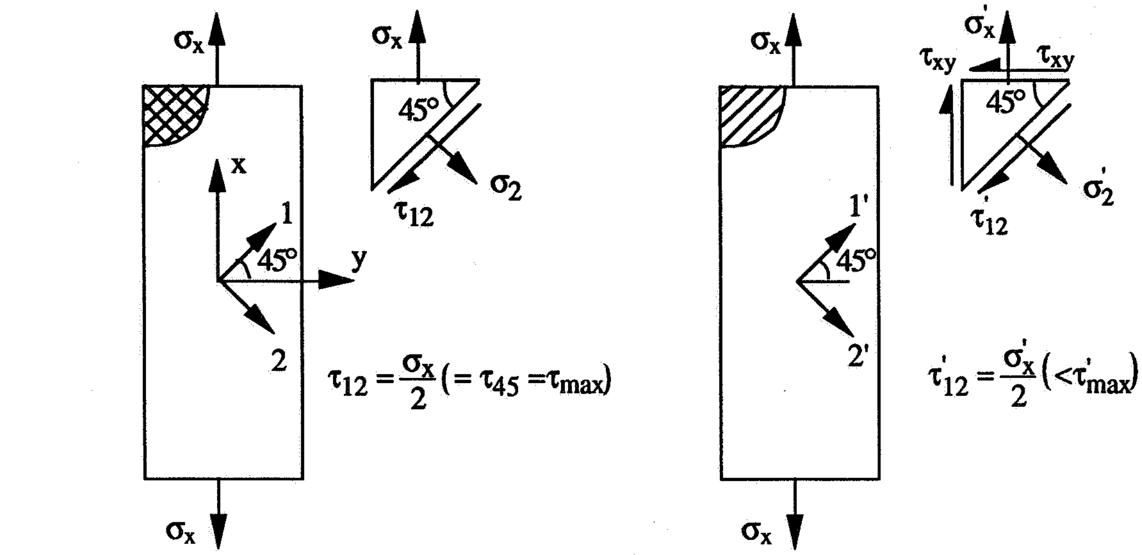


Fig.18 Normalized strength (top) and normalized strain at failure (bottom) versus specimen size for epoxy specimens scaled in three dimensions. For specimens with distributed plies an apparent strength or strain at failure value is used, corresponding to the point where the gradient of the stress/strain curve is zero, see figure 3.



(a) Mohr's Circle for **Laminate** Stresses (b) Mohr's Circle for **Ply** Stresses

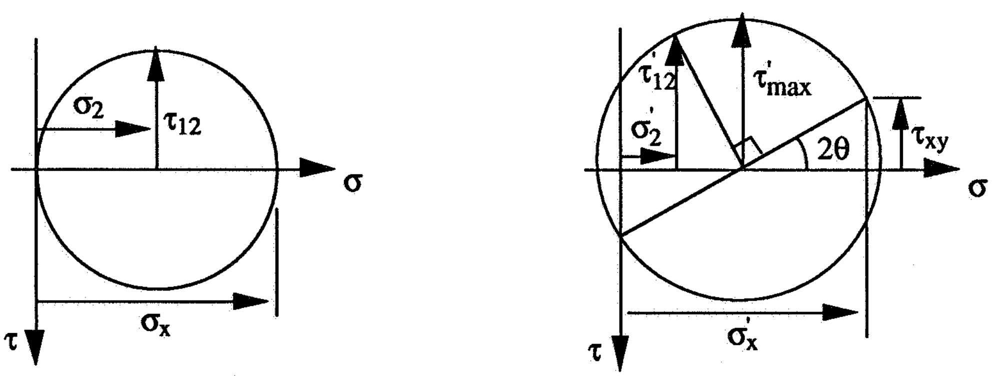


Fig. 19 The relationship between the applied and local stresses at laminate and ply level. Plane and uniform stress conditions have been assumed.

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Sotiris Kellas and John Morton

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Virginia Polytechnic Institute and State University
Dept. of Engineering Science and Mechanics
225 Norris Hall
Blacksburg, VA 24061

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13. ABSTRACT *(Maximum 200 words)*

The effect of specimen size upon the response and strength of ±45° angle-ply laminates has been investigated for two graphite fiber reinforced plastic systems and several stacking sequences. The first material system was a brittle epoxy based system, AS4 fibers in 3502 epoxy, and the second was a tough thermoplastic based system, AS4 fibers in PEEK matrix. For the epoxy based system two generic ±45° lay-ups were studied; $(+45°_n/-45°_n)_{2S}$ (blocked plies), and $(+45°/-45°)_{2nS}$, for n=1 and 2. The in-plane dimensions of the specimens were varied such that the width/length relationship was 12.7xn/127xn mm, for n=1, 2, 3, or 4.

It is shown that the stress/strain response and the ultimate strength of these angle-ply laminates depends on the laminate thickness and the type of generic lay-up used. Furthermore, it is shown that first ply failure occurs in the surface plies as a result of normal rather than shear stresses. The implications of the experimental findings upon the validity of the ±45° tensile test which is used to determine the in-plane shear response of unidirectional composites are discussed.

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