V. FAILURE OF FIBER COMPOSITE LAMINATES – PROGRESSIVE DAMAGE AND POLYNOMIAL INVARIANTS

Fiber composites are typically used in flat, laminated forms or variations thereof, often also involving woven or braided constructions. The treatment given here will be for the flat laminated form where the aligned fibers in the various lamina take different directions within the laminate. Failure at the laminate level is complex and not an obvious matter in terms of deducing the theoretical basis for the related failure criteria. Two very different but mainline methods will be given here, although there are many other approaches.

The best place to start is with an understanding of the defect states that can and do exist in commercial fiber composite products. These defects have a profound influence upon the strength performance of the resulting materials. There are broken fibers, fiber slacks, fiber misalignments, debonds, resin rich pockets, cracks, porosity, and on and on. Effective moduli do not depend much on the defects, but failure surely does.

With all the defects, one could question how any composite material could perform at a satisfactory level. Despite the many defects, the special properties of the fibers and the ameliorating role of the matrix combine to give an extraordinary balance of properties that remains at a very high level for the higher quality products. To be sure, these composite properties would not be possible without a superior capability of the fiber phase or the enabling properties of the matrix phase, both are vital and essential.

Next on the scale of importance is the matter of scale itself. One of the major fulcrums of dissension is the uncertainty as to the scale at which to characterize failure. All failure modes have a scale of action. The central problem is to recognize the scale for a particular mode of failure and then to characterize it. Damage occurs at all scales from molecular on up.

Perhaps an extreme example will best illustrate the point. Molecular dynamics cannot predict the Euler buckling of a main structural member. This mismatch in scales is ridiculously great. While it is generally taken that smaller scales reveal more fundamental effects than do larger scales, there are important and critical exceptions to this oversimplification. Many or most materials composing load bearing structures fall under the exceptional category. These engineering materials are strongly distinguished from other major classes such as electronic materials and bio-materials which have far different primary functions from that of mainly supporting load without failure but with minimal deformation.

Composites laminates have failure modes that inherently relate to the scale of the laminate. For example, delamination is meaningful only at the laminate scale. The different but very broad scales of conceivable relevance here are

Atomic – angstroms 10^{-9} m
Fiber size – microns 10^{-6} m
Lamina thickness – mils 10^{-4} m
Laminate thickness – millimeters to inches 10^{-1} m

Concerning scale dependence, kink bands for example cannot be predicted from fiber scale failure. More generally, contrary to many claims, macroscopic failure cannot be predicted by molecular dynamics. There is a 6 or 7 orders of magnitude gap that is an imposing barrier. Realistic macroscopic failure predictions must originate from a careful assessment of the proper scale for the failure mode. Many, but not necessarily all macroscopic failure modes originate at the macroscopic scale. Great care is needed in making this determination. Taking macroscopic failure to be at the lamina and laminate scales, the origin of the failure modes then depends upon the particulars of the layup of the lamina within the laminate.

Although interesting as concepts and certainly useful in developing new materials, truly using nano-mechanics and/or micromechanics effects to predict macroscopic failure would require an immense commitment and by no means is it an established approach. Some micro or nano scale approaches are that in name only since they “back out” constituent properties to give desired macroscopic scale results. Infact, even just trying to predict laminate failure behavior from lamina level behavior may have severe limitations, as will be shown later. Due to the complications of damage and of the scale dependence of the failure modes, failure characterization based upon nano-scale and microscale idealizations (either fully or partially so) will not be considered in the two methods to be given here.

There is a very large literature on the subject of damage and failure in fiber composite laminates. All conceivable approaches are being tried, a few typical approaches will be mentioned here. Hinton, Kaddour and Soden[1] complied a very useful collection and assessment of many different approaches. Puck and Schurmann [2] give an approach originally based upon Coulomb-Mohr type behavior. Mayes and Hansen [3] descriptively designate their method as MCT (Multi-Continuum Theory). Daniel [4] treats many aspects of failure. Robbins and Reddy [5] give an approach using internal variables. Tsai and colleagues [6] give and use MMF (Micromechanics of Failure). A finite element approach to damage and failure is given by Tay et al [7], EFM (Element Failure Method). Not only are there many research sources and references, almost all of them represent distinctly different and individualized approaches. Considering the difficulty of the topic, it is not surprising that there is nothing that is even within proximity of being a unified, verified, and reasonably recognized methodology.

The present interests and methods will be restricted to carbon fiber–polymeric matrix systems, which are generally the highest performance systems. As such, these approaches may be somewhat less applicable to glass – polymer systems, although the possibility remains open.

The technical starting point is to recall the lamina level treatment given in Section III. Typical properties for aligned carbon fiber-epoxy matrix lamina are given by

$$
\begin{aligned}
E_{11} &= 150 \text{ GPa} \\
E_{22} &= 9 \text{ GPa} \\
\mu_{12} &= 6 \text{ GPa} \\
\mu_{23} &= 3 \text{ GPa} \\
\nu_{12} &= 1/3 \\
\nu_{23} &= 1/2
\end{aligned}
$$

and

$$
\begin{aligned}
T_{11} &= 2000 \text{ MPa} \\
C_{11} &= 1500 \text{ MPa} \\
T_{22} &= 40 \text{ MPa} \\
C_{22} &= 150 \text{ MPa} \\
S_{12} &= 80 \text{ MPa} \\
S_{23} &= 50 \text{ MPa}
\end{aligned}
\qquad (1)
$$

The two independent lamina level failure criteria derived in Section III are given by

Fiber Controlled Failure:

$$\left(\frac{1}{T_{11}}-\frac{1}{C_{11}}\right)\sigma_{11}+\frac{1}{T_{11}C_{11}}\sigma_{11}^{2}\leq 1 \qquad \text{(2a)}$$

or simply

$$-C_{11}\leq\ \sigma_{11}\leq T_{11} \qquad \text{(2b)}$$

Matrix Controlled Failure:

$$\left(\frac{1}{T_{22}}-\frac{1}{C_{22}}\right)\left(\sigma_{22}+\sigma_{33}\right)+\frac{1}{T_{22}C_{22}}\left(\sigma_{22}+\sigma_{33}\right)^{2}+$$

$$\frac{1}{S_{23}^{2}}\left(\sigma_{23}^{2}-\sigma_{22}\sigma_{33}\right)+\frac{1}{S_{12}^{2}}\left(\sigma_{12}^{2}+\sigma_{31}^{2}\right)\leq 1 \qquad \text{(3)}$$

There is something especially attractive about a lamina level failure criterion that decomposes into separate fiber controlled and matrix controlled failure modes. They are not the obvious approach of explicitly treating fiber failure as distinct from matrix failure and conversely treating matrix failure as distinct from fiber failure. They do allow interactive fiber and matrix effects but with all the defects included in an implicit way, and still they decompose into the two separated modes of failure. Two other lamina level failure criteria are given in Section III, namely the Tsai-Wu and the Hashin forms. The first does not allow the decomposition of failure modes while the second involves four different decomposed sub-forms.

Concerning the uniaxial lamina failure criteria given above, it is seen that the stress in the fiber direction is uncoupled from the other stress components. Should all of these stress components be coupled? One could assume a coupled form and then evaluate the coupling parameters from failure data. While these would then fit a particular set of data, they could hardly be considered to be general properties of failure. It is here preferred to use the theoretically derived and idealized form specified by only the explicit failure properties in the above criterion, and by nothing else. This then is considered to give the essential elements of failure at the lamina scale. This is what will be used in approaching the method of progressive damage at the laminate scale. Actually, this method bridges between the lamina and laminate scales.

Progressive Damage

The idea behind progressive damage is quite simple. Both matrix controlled and fiber controlled types of failure can separately and sequentially occur during the loading of the various lamina within a laminate. At some point so much damage has accumulated in the form of these local failures that the laminate can no longer sustain load. This then comprises the ultimate load, failure in the broad and total sense.

So progressive damage in laminates is not so much an original and stand alone discipline as it is a careful accounting of the sequence of local failures (damage) leading up to the complete failure of the laminate. The lamina level failure criteria for the fiber controlled and the matrix controlled modes still must be selected from a variety of existing forms. Thus the term progressive damage essentially means predicting laminate failure from lamina level damage and failure.

The standard, direct way of proceeding places high priority on modeling the failure at the lamina level and then using that to build up to laminate failure. The presumption then is that high accuracy at the lamina level would translate into highly reliable results at the laminate level. This is the conventional thinking. While this line is quite reasonable on its face, the problem is in what constitutes high accuracy at the lamina level and how does that translate to the laminate level.

In the immediately following developments, the in-plane and the out- of-plane failure modes for the laminate will be taken to be uncoupled due to the extreme anisotropy of effects between them. First, the in-planes cases will be examined in considerable detail. Later the out-of-plane (delamination) cases will be taken up.

Begin with failure of the quasi-isotropic layup case. The common form for this involves equal numbers of lamina in the 0 , +45, -45, and 90 degree fiber directions. The failure of the quasi-isotropic layup is probably the most important single case that can be studied. This limiting case is one of the two anchors of all possible layups. The other limiting case is the unidirectional form itself. All other cases lie between these two extremes. Thus the quasi-isotropic case is one of the most severe tests of using a lamina level failure criterion to predict damage and failure of a laminate.

The first damage that usually occurs is that of the matrix failure in some lamina. But this is not of as much interest here as is the complete failure involving the fiber load carrying capability. Taking the properties and lamina level failure criteria given above, the exact sequence of failures can be determined for the various lamina within the quasi-isotropic laminate. These are given by the following sequences for some particularly important stress states. The unidirectional stress states are aligned with the 0 degree fiber direction. Lamina level failure criteria (2) and (3) along with properties (1) are used in the progressive damage analysis.

Uniaxial Tension

(i) 90˚ Matrix damage, then
(ii) ±45˚ Matrix damage, then
(iii) 0˚ Fiber failure

Uniaxial Compression

(i) ±45˚ Matrix damage, simultaneous with
(i) 0˚ Fiber failure

Eqi-Biaxial Tension

(i) All matrix damage, then
(ii) All fiber failure

Eqi-Biaxial Compression

(i) All fiber failure, no matrix damage

Shear (0˚ Ten, 90˚ Comp)

(i) ±45˚ Matrix damage, then
(ii) 90˚ Matrix damage, then
(iii) 90˚ Fiber failure

In these examples the corresponding moduli are set to zero after the failure criteria prescribes the matrix controlled or fiber controlled failure. The fiber failures shown lead directly to overloading the other lamina and total failure ensues. The details of this accounting approach are lengthy, but completely routine.

When the failure mode is matrix controlled, the resulting “jump” in strain at constant stress is small compared with the jump in strain that occurs when the fiber mode fails in a particular lamina. This motivates a simplification that is here designated as “fiber dominated” progressive damage. Again, this is appropriate mainly to carbon-polymer systems and it preserves the general shape of the failure envelopes.

Henceforth in the progressive damage examples the fiber dominated state will be used whereby the matrix controlled properties are taken to be negligible compared with the fiber controlled properties. Then only a sequence of fiber controlled failures in the various lamina need be considered.

For the present properties example, and for the quasi-isotropic laminate, the fiber dominated biaxial stress failure envelope is given in Fig. 1 for the case of T=C. In this case subscripts are not needed on the in-plane uniaxial strengths. Tensile and compressive strengths of 2000 MPa are used for the fiber direction lamina strengths.

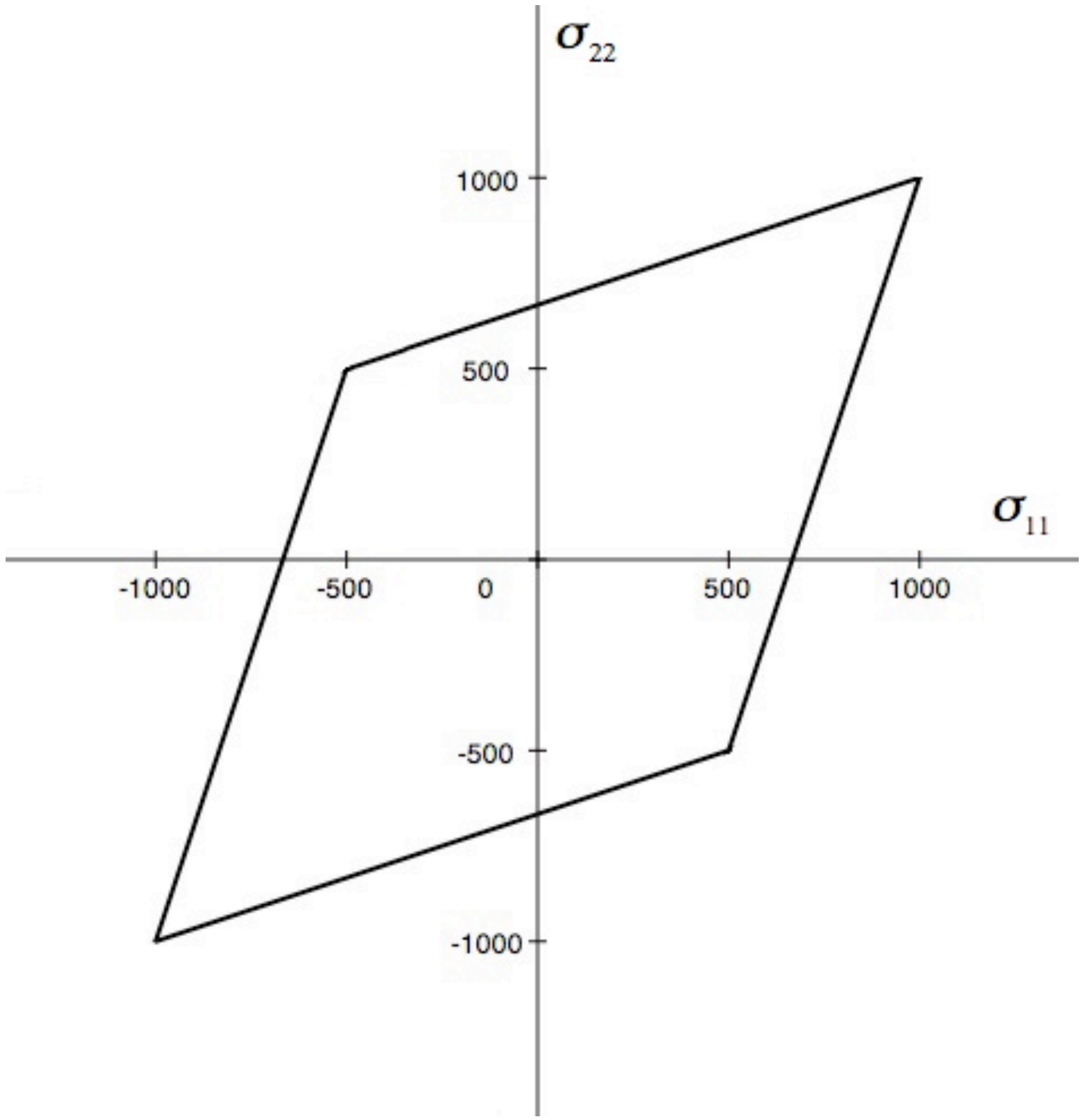

Fig. 1 Quasi-Isotropic failure envelope, progressive damage, T=C

In Fig. 1 and from this point on, the 1, 2, 3 coordinate system refers to the laminate with the 3 axis in the thickness direction. It is seen that the failure envelope takes a diamond shaped form There is reflection symmetry about axes that are rotated 45 degrees to those in Fig. 1.

The failure envelope due to progressive damage in the quasi-isotropic laminate with fiber direction strengths from (1) is shown in Fig. 2.

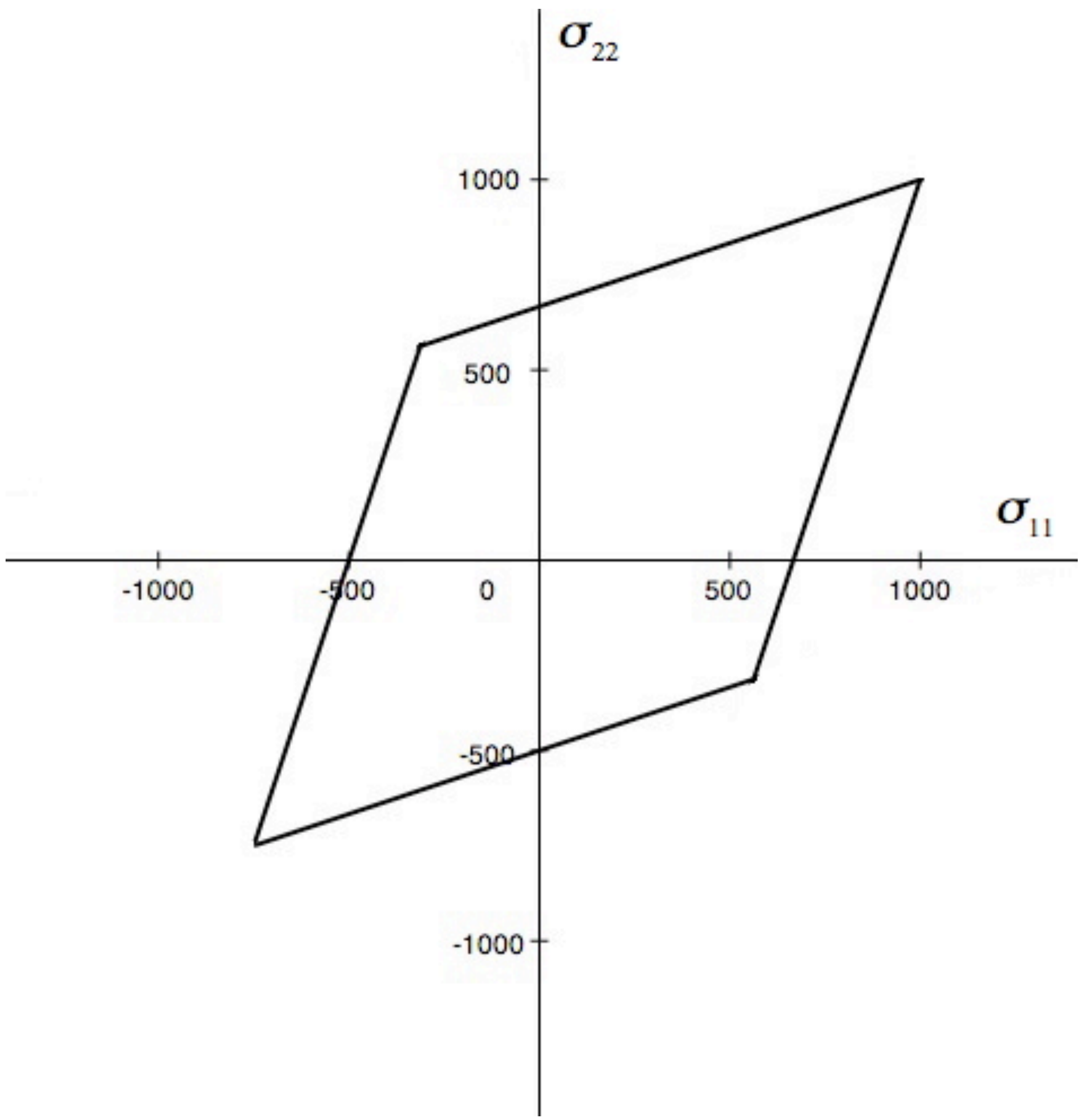

Fig. 2 Quasi-Isotropic failure envelope, progressive damage, T≠C

The T≠C case has a lower degree of symmetry than that for T=C. The diamond shape persists and is characteristic of the progressive damage approach.

Now an orthotropic example will be given. Take a layup that has a strong emphasis on the reinforcement in a particular direction, with the volume fractions

$$
\begin{array}{ll}
0^\circ \ \text{lamina}, & c=1/2 \\
+45^\circ \ \text{lamina}, & c=1/6 \\
-45^\circ \ \text{lamina}, & c=1/6 \\
90^\circ \ \text{lamina}, & c=1/6
\end{array}
\qquad (4)
$$

Using the fiber dominated properties from (1) gives the failure envelope as

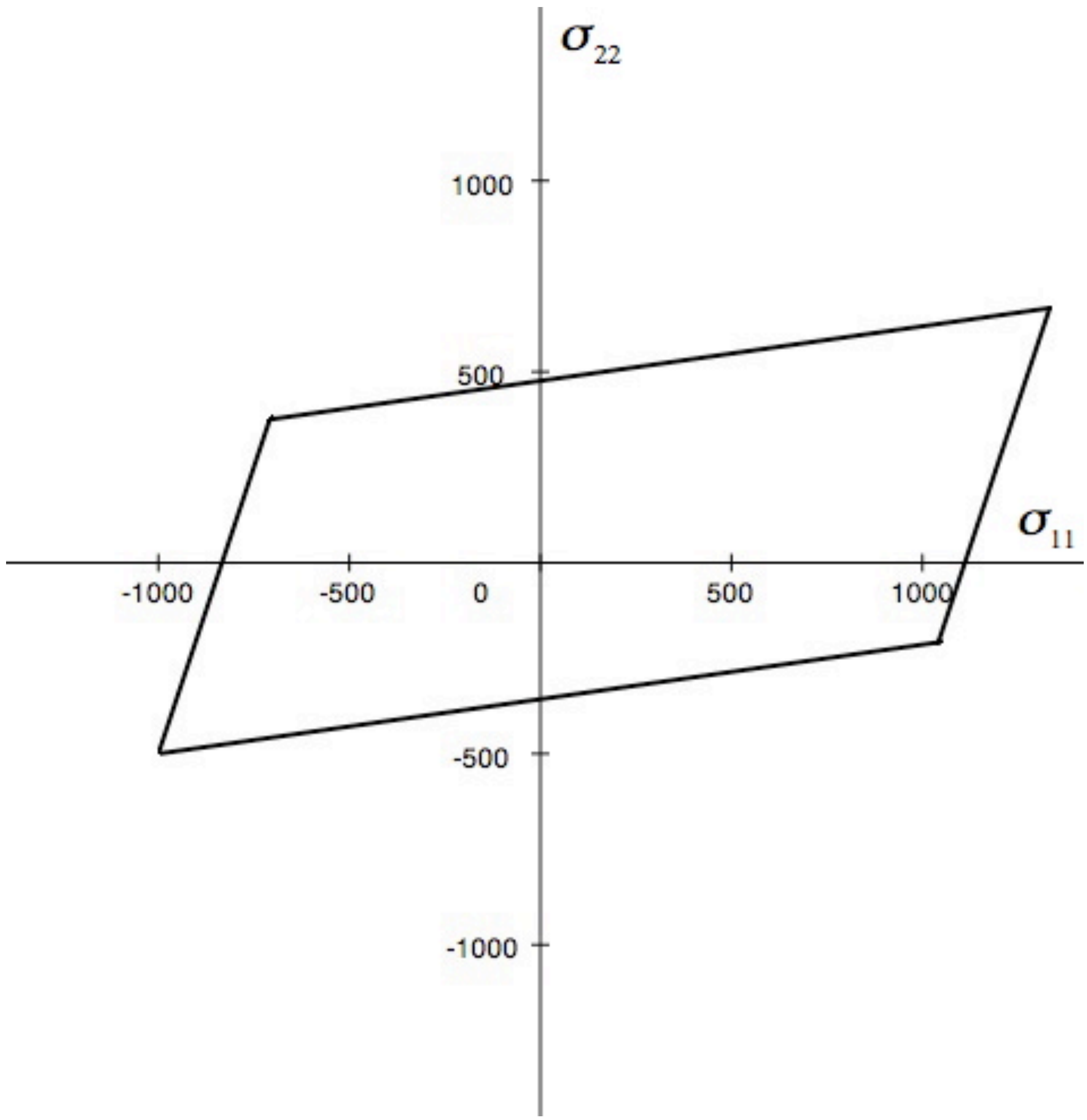

Fig. 3 Orthotropic failure envelope, progressive damage

As seen from Fig. 3 the degree of anisotropy in the strengths is larger than 2 to 1.

These diamond shaped failure envelopes are like those of the maximum normal strain criterion for three dimensionally isotropic materials. This similarity is not just a coincidence. At the lamina level the failure criterion is that of stress in the fiber direction. For a high degree of anisotropy at the lamina level, the criterion of stress in the fiber direction is well approximated by strain in the fiber direction. This interpretation is useful because strain in the fiber direction at the lamina level is the same as strain in a fiber direction at the laminate level, for the in-plane conditions. Thus in these examples the laminate failure criterion is that of maximum strain in a (any) fiber direction.

The progressive damage behavior shown in Figs. 1-3 is typical for most layup forms. These results are somewhat surprising, especially as regards the acute angles of the failure envelopes in the first and third quadrants and the associated very large allowable stresses. The obvious question is whether these acute angles reflect expected physical behavior, or are they artifacts of the method of progressive damage.

Testing Results

Hinton, Kaddour, and Soden[1] conducted a lengthy and detailed evaluation of many different fiber composite failure theories. Most of the theories that were included in their study involved variations of progressive damage. In particular for a quasi-isotropic laminate they found rather good agreement between data which they cited (by Swanson) and progressive damage in the first, second, and fourth quadrants of biaxial stress states. It should be noted that the data in the first quadrant did not extend all the way to the apex of the predicted failure locus. They did not find good agreement in the third quadrant, but that was attributed to an experimental deficiency, the buckling of the specimens in compression-compression stress states.

There is a different source of data that gives a different conclusion from that mentioned above. Specifically, data generated and reported by Welsh, Mayes and Biskner [8] does not support the first and third quadrant predictions of progressive damage for carbon-epoxy, quasi-isotropic laminates in biaxial stress states. Infact, they reported that their data are fairly well described by a Mises-like form with no corners at all in the failure envelope.

The Welsh, Mayes, Biskner data are shown in Fig. 4, using symmetry about a 45 degree line through the origin.

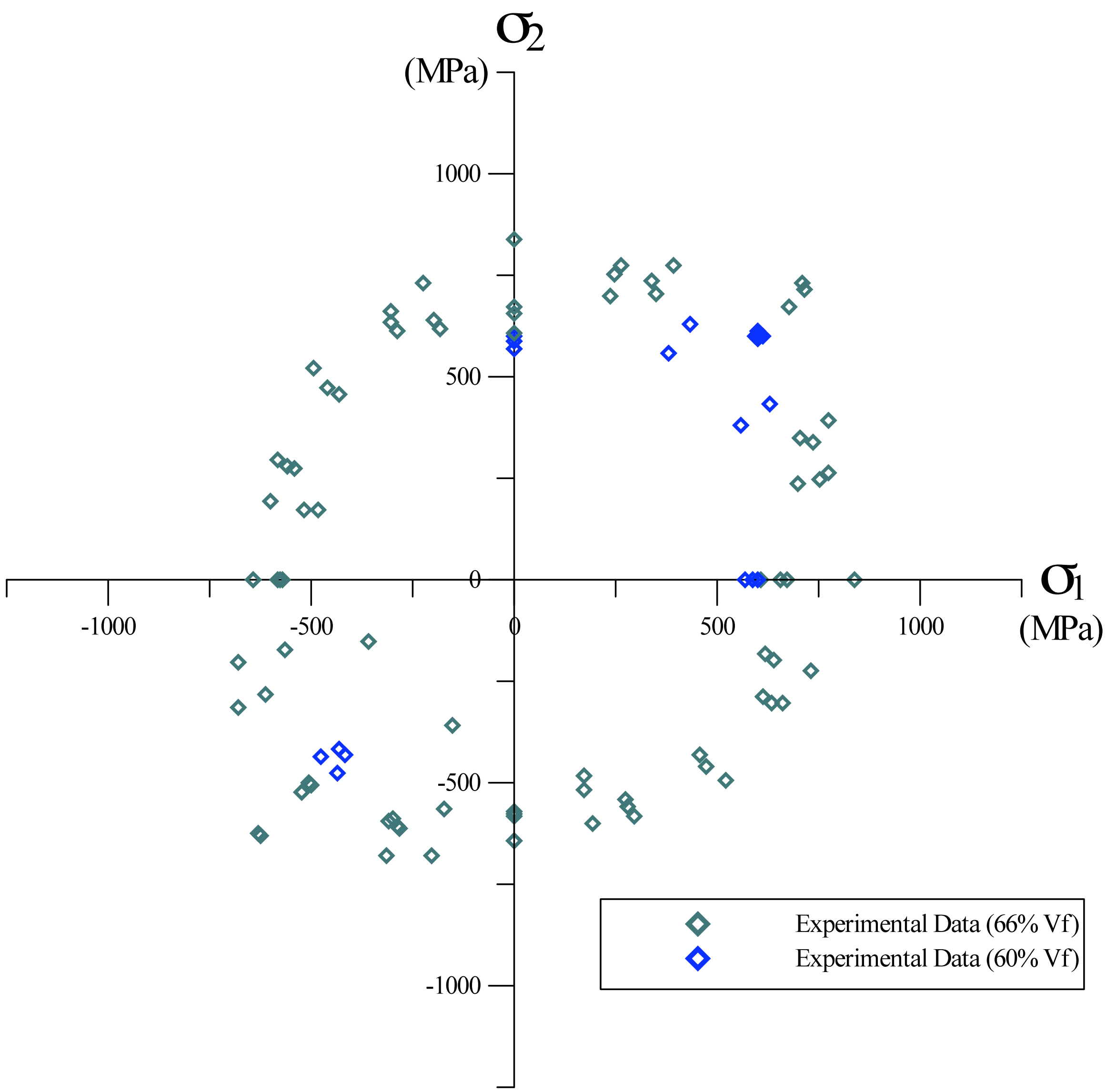

Fig. 4 Quasi-Isotropic laminate failure data, IM7/977-2
(courtesy of Dr. J. S. Welsh)

Thus there are conflicting data sets for the failure behavior of quasi-isotropic laminates. One data set supports the existence of the acute angle corners in the biaxial failure envelope while another does not support it.

From a theoretical point of view, the form with corners is prescribed by the maximum strain type of failure criterion (in a fiber direction). The maximum strain criterion has no credibility for isotropic materials such as metals and polymers, but despite this, does it and should it apply for fiber composite laminates? The entire situation is somewhat ambiguous and certainly unanswered.

Perhaps the strain criterion in the fiber direction is too simplistic for laminates. It is a one dimensional criterion. Perhaps laminates, as distinct from a lamina within the laminate, requires a failure criterion that allows more than one dimensional stress and strain representations. Perhaps laminates allow more critical (lower level) failure modes than those permitted by one dimensional strain. The genesis for the one dimensions strain criterion is likely the overly simplified view that composites are composed of perfectly straight fibers that are perfectly collimated and spaced in an ideal matrix phase.

It is seen that this progressive damage approach is built upon a set of conjectures as to what the failure mode should be. However, a qualification should be added to this. It is best viewed as what the failure mode "could be", not "should be". It may be best to view this standard progressive damage approach as an upper bound. It is a plausible failure mode but it could be superceded by a different failure mode that operates at a different scale and at a lower critical value of stress in the laminate.

Another complication is that of insitu properties and damage. In quasi-isotropic laminates under uniaxial tensile stress the lamina under the most load in the fiber direction has no significant transverse stresses causing matrix cracking. But in eqi-biaxial tension all lamina have very extensive matrix cracking and damage. The transverse cracking also can further degenerate into delamination. In eqi-biaxial tension the damage is invasive and it compromises the load carrying capacity in the fiber direction within any lamina. The diamond shaped failure envelope from progressive damage is too idealized to accurately reflect such degrading effects. A more conservative and more physically realistic failure envelope may be called for.

Polynomial Invariants

If there were a need to characterize failure for a given laminate configuration and the method of progressive damage was not available, how would one approach the problem? Almost certainly one would use the same general method as was used in Section II for isotropic materials. Specifically, one would take a polynomial expansion in the stress invariants for the symmetry of interest. Then truncate the expansion at second degree terms. As then expressed, the form would involve specific constants or parameters to be interpreted as the failure properties. By obvious description the method should be designated as that of "Polynomial Invariants". Exactly this method will be developed and followed here for the failure of laminates.

The reason why the polynomial invariants method may be more appropriate than progressive damage is that some failures are at the laminate level, not the lamina level. Thus some failure modes involve cooperative motions between the lamina within the laminate. Any failure mode has a scale of action at which it is relevant, and the laminate failure modes should be considered.

The first developments and examples will be for quasi-isotropic laminates. The failure modes will be concerned with the in-plane ones and then later the out-of-plane ones (delamination) will be developed. The term quasi-isotropic refers to in-plane elastic moduli (or compliance) properties. It is an entirely separate question to consider whether the failure properties are also quasi-isotropic. A close examination of the progressive damage method shows that its strength properties are not quasi-istropic even when the moduli are. In contrast, for the polynomial invariants method to be developed here, the failure properties will be taken to be quasi-isotropic when the laminate has quasi-isotropic moduli.

The in-plane invariants for quasi-isotropic behavior are

$$\left(\sigma_{11}+\sigma_{22}\right) \quad and \quad \left(\sigma_{12}^2-\sigma_{11}\sigma_{22}\right)$$

Taking the polynomial expansion in these invariants and truncating at second degree terms gives the failure criterion as

$$\left(\frac{1}{T}-\frac{1}{C}\right)\left(\sigma_{11}+\sigma_{22}\right)+\frac{1}{TC}\left(\sigma_{11}+\sigma_{22}\right)^2+ \frac{1}{S^2}\left(\sigma_{12}^2-\sigma_{11}\sigma_{22}\right)\leq 1 \tag{5}$$

where T and C are the in-plane uniaxial tensile and compressive strengths and S is the in-plane shear strength.

Many or perhaps most fiber composites applications involve orthotropy. For the case of orthotropy, the full three dimensional form for polynomial invariants involves 12 independent properties. When specialized to the case of fiber composite laminates, with the in-plane and out-of-plane forms being decomposed, the resulting polynomial invariants form for the in-plane case involves 6 properties and is given by

$$\left(\frac{1}{T_{11}}-\frac{1}{C_{11}}\right)\sigma_{11}+\frac{1}{T_{11}C_{11}}\sigma_{11}^{2}+$$
$$\left(\frac{1}{T_{22}}-\frac{1}{C_{22}}\right)\sigma_{22}+\frac{1}{T_{22}C_{22}}\sigma_{22}^{2}+$$
$$\lambda_{12}\sigma_{11}\sigma_{22}+\frac{\sigma_{12}^{2}}{S_{12}^{2}}\leq 1 \qquad (6)$$

where the T's and C's and S_{12} have the obvious strength identifications, but λ_{12} is yet another independent strength property.

These two expressions are the final results from the polynomial invariants method for the quasi-isotropic and the orthotropic cases. Three strength properties are required for the former case and six for the latter. These results are not subject to the fiber dominated condition used in the preceding development of progressive damage.

For purposes of comparing these results with those of fiber dominated progressive damage, the numbers of independent strength properties in (5) and (6) will be reduced somewhat. First for the quasi-isotropic case, in the special case where T=C, for the polynomial invariants method to give the same slope as that of progressive damage where the envelope crosses the axes then determines S in terms of T as

$$S^2 = \frac{3}{8}T^2 \tag{7}$$

Now in the case where T≠C, the similar procedure gives the best fit between the two methods as

$$S^2 = \frac{3}{8}TC \tag{8}$$

In the orthotropic case, the property λ_2 can be determined (not uniquely) in terms of the T's and C's as

$$\lambda_{12} = -\frac{2}{3\sqrt{T_{11}C_{11}T_{22}C_{22}}} \tag{9}$$

where the 2/3 factor is found by requiring that the form (9) reduce to the proper result in the quasi-isotropic specialization of the orthotropic form.

A fiber dominated estimate of strength property S_{12} in terms of the usual known properties is given by

$$S_{12} = \sqrt{\frac{T_x C_x}{6}}\left[1-\left(c_0 + c_{90}\right)\right] \tag{10}$$

where T_x and C_x are the lamina level fiber direction strengths and c_0 and c_{90} are the volume fractions of the 0 and 90 degree lamina in the orthotropic laminate.

With relation (8) the properties needed for the quasi-isotropic laminate reduces to only T and C. With relations (9) and (10) the properties needed for the orthotropic laminate reduces to only T_{11}, C_{11}, T_{22}, and C_{22}. Of course it would be far preferable to experimentally determine all three properties in the quasi-isotropic case and all six properties in the orthotropic case, but there are situations where this is difficult and only the uniaxial tests can readily be performed.

Fig 5 shows the polynomial invariants, quasi-isotropic prediction for T=C and Fig. 6 for T≠C, both from (5) and using (7) and (8) respectively. Also shown are the predictions from progressive damage, with both methods calibrated to give the same uniaxial strengths in the two directions.

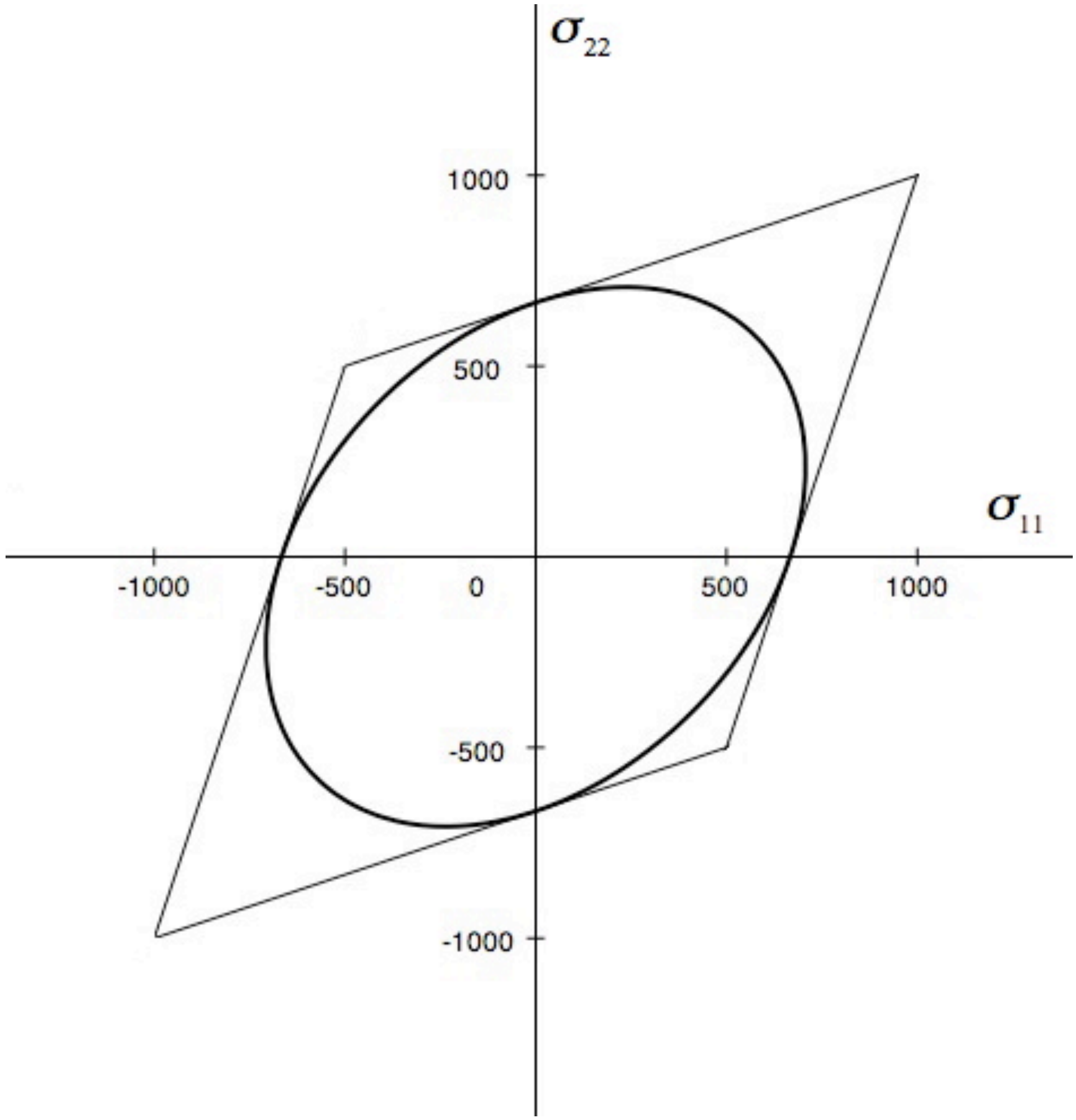

Fig. 5 Quasi-Isotropic failure, polynomial invariants, T=C

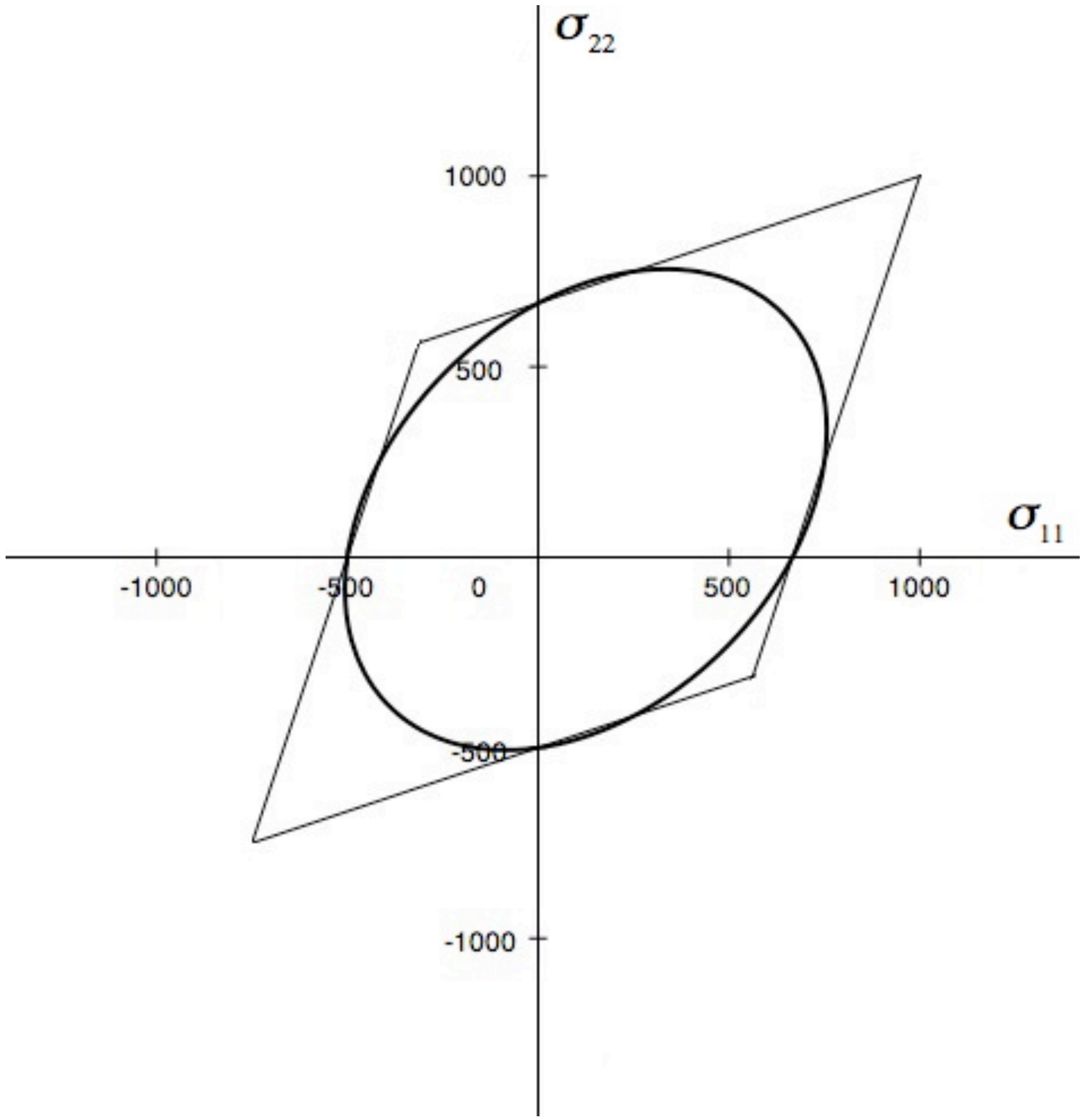

Fig. 6 Quasi-Isotropic failure, polynomial invariants, T≠C

It is now apparent that the polynomial invariants prediction would give a more realistic model of the data in Fig. 4 than would progressive damage.

Fig. 7 shows the orthotropic laminate example failure prediction from polynomial invariants, (6) along with (9). This orthotropic laminate has the layup pattern specified earlier, (4).

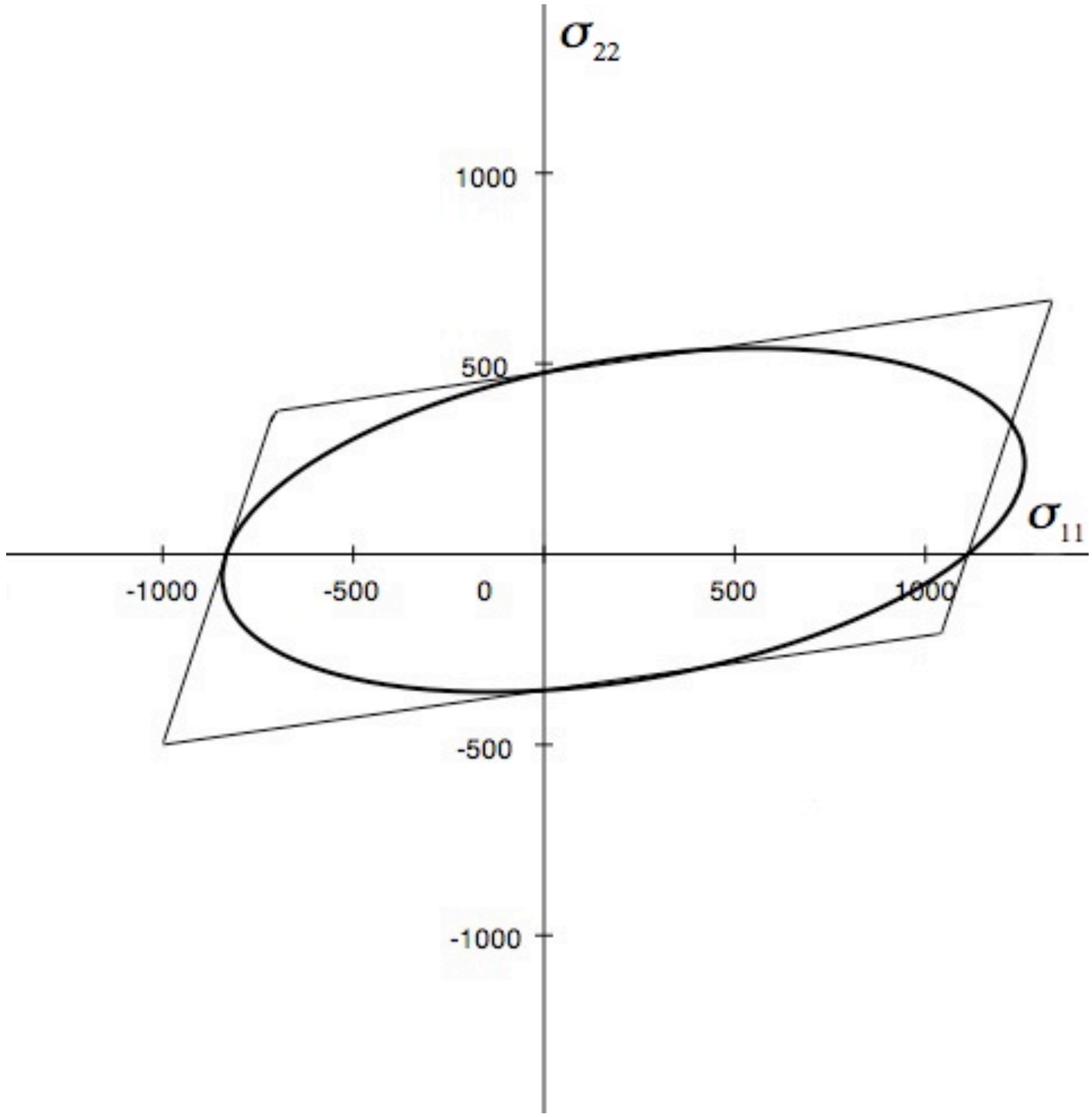

Fig. 7 Orthotropic failure, polynomial invariants

The comparisons and differences between polynomial invariants and progressive damage predictions are striking. The polynomial invariants approach has the rather "smooth" behavior often seen when complicated physical effects along with a great many defects and imperfections are at interaction. Progressive damage appears to be a much more idealized and in effect non-conservative prediction of behavior. In another sense the difference is perhaps in even sharper focus. Progressive damage predicts failure at the lamina level, and presumes that it controls laminate behavior.

In contrast, polynomial invariants predicts and is based only upon failure modes at the laminate scale. Which method is most reflective of physical reality? That remains as an open and likely contentious question. The limited data that are available are contradictory on this question.

It should be remembered that the progressive damage predictions are based upon the theoretically derived lamina level failure criteria. If one abandons that particular basis and simply adjusts lamina level failure envelopes then just about any desired laminate level prediction could be obtained through progressive damage. While that might satisfy a particular set of data, there could be no confidence in the generality of that approach. The view favored here is that the polynomial invariants method has a more solid grounding and it is more likely to be generally useful. These questions and uncertainties will be taken up in the future section concerned with critical tests for failure criteria.

There remains the possibility that for preliminary design purposes, progressive damage could be used to give the uniaxial strengths for any particular laminate configuration and then that is used in the polynomial invariants method to give the complete failure envelope.

A note on terminology may be helpful. The term polynomial invariants that is developed and applied here for qausi-isotropic and orthotropic laminates of course could be applied for any symmetry of interest. Infact the isotropic material case of Section II certainly employed polynomial invariants, although this terminology was not explicitly used there. The three lamina level failure criteria, (2)&(3) and the Tsai-Wu and the Hashin forms, also represent special cases of polynomial invariants.

Up to this point, only the in-plane failure characterization has been considered. Now the out-of-plane, delamination, case will be addressed. Using the method of polynomial invariants, it is quickly found that the failure criterion for delamination is given by

$$\left(\frac{1}{T_{33}}-\frac{1}{C_{33}}\right)\sigma_{33}+\frac{1}{T_{33}C_{33}}\sigma_{33}^2+\frac{1}{S_{23}^2}\left(\sigma_{23}^2+\sigma_{31}^2\right)\leq 1 \tag{11}$$

where T_{33} and C_{33} are the thickness direction uniaxial strengths and S_{23} is the interlaminar shear strength. The form (11) is the delamination failure criterion (assumed) for all layups, and the stress components are those transmitted across the interfaces.

Generally but not necessarily the failure in the thickness direction is due to delamination between laminae rather than failure within the individual lamina themselves. It is treated as such here. The delamination failure is essentially that of the failure of the extremely thin polymer layer of locally fluctuating thickness between the lamina. Combinations of through the thickness tension and interlaminar shear are seen to be especially limiting in (11).

The failure criterion (11) is a fail/no fail type of overall criterion. Sometimes delamination is of a progressing and advancing nature. This type of behavior can be effectively treated using fracture mechanics with a cohesive zone failure model.

In many circumstances the "weak links" with fiber composites are the transverse tensile cracking of the matrix phase and that of delamination. If no matrix cracking is to be allowed then the matrix controlled failure criterion (3) at the lamina level can be used in design to prevent it. Often however some small degree transverse matrix cracking can be tolerated as containable damage. The situation with delamination is usually much more serious and no delamination at all can be allowed. Any area of delamination destroys the compatibility between the lamina and renders the laminate as disfunctional. Failure criterion (11) (or some alternative form) must be used to prevent delamination from happening under ordinary conditions.

The in-plane and the out-of-plane failure modes have thus far been taken to be independent. In reality there could be some coupling between them. Infact, transverse cracking in the lamina can nucleate delamination in the laminate. It is know that in-plane failure can, in some cases, involve complex combinations of partial delamination with the fiber controlled failure. The direct method of progressive damage shown here cannot account for such interactive effects. In contrast, the method of polynomial invariants, as applied to the entire laminate, can and does implicitly contain all such effects, including the effects of the lamina stacking sequence.

Further complications which can occur are those of interlaminar edge effects in the testing of laminates. Such effects can obscure the overall objective of determining the intrinsic mechanical properties of failure for the laminate. It is seen that many of the complications that arise with the failure of laminates arise only at the explicit scale of the laminate and are completely invisible to failure characterization at smaller scales. Progressive damage treats laminate failure as superimposed failure effects from those at smaller scales. Failure is of a completely nonlinear character and must include all effects operative up to and including the scale of interest.

It should not be construed that the proper failure criterion for laminates would obviate the need for a proper failure criterion at the lamina level. Both would be equal partners in understanding all failure effects. Indeed, the individual fiber and matrix failure criteria and capaility levels are also of great interest as independent matters. Forms for three of these four categories are given in this website.

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