IS THERE REALLY NO NEED TO BE ABLE TO PREDICT MATRIX FAILURES IN FIBRE-POLYMER COMPOSITE STRUCTURES?
PART 1: EXPLANATION OF FATAL FLAWS IN EXISTING THEORIES

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IS THERE REALLY NO NEED TO BE ABLE TO PREDICT MATRIX FAILURES IN FIBRE-POLYMER COMPOSITE STRUCTURES?
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ABSTRACT

A logic-based case is presented that it IS necessary to be able to predict matrix failures in fibre-polymer composites, even though some 50 years after the introduction of the advanced fibre-polymer composites there is still no widely accepted theory capable of actually doing so. This lack is due to the widespread malpractice of artificially homogenising distinct fibre and matrix constituents into an “equivalent” anisotropic solid, to simplify the mathematics. The enormous residual thermal stresses in the matrix, even when the lamina is stress-free, are not widely recognised; these detract appreciably from the ability of the matrix to withstand applied mechanical loads. The extreme thermal mismatch between the fibres and the polymer causes the resin to try to shrink during cool-down after cure at high temperature, but the strong stiff fibres prevent this from happening. Such internal residual stresses cannot possibly exist in truly homogeneous materials, so the consequences are overlooked whenever homogenised material models are used in analyses when they shouldn’t have been. The case is made that it is impossible to predict matrix failures using any theory in which the fibre and resin constituents have been artificially homogenised, despite the myriad of theories claiming to do so that can’t.

Keywords: fibre-polymer composites, matrix failures, internal residual thermal stresses, bogus interactive failure models, false assumption of homogeneity, SIFT failure model

INTRODUCTION

Possibly because the diameters of fibres in high-strength fibre-reinforced plastics are extremely small, a great many structural analysts and designers have come to think of fibre-polymer composites as homogeneous anisotropic solids, rather than as two-constituent materials containing discrete fibre and matrix phases. This over-simplification has never been validated. Indeed, it is so deeply ingrained in composite analysis methods nowadays that few researchers even question its validity. It can be validated only by deriving a better theory that yields the same predicted strengths with or without the simplifying assumption. It can no longer be validated by experimental data because the widespread acceptance of composite failure theories using theories with adjustable parameters to replace true material properties, without requiring that the theories be applied only with the measured properties, has discredited this approach to the point that it is no longer credible. Instead of recognising that the need to adjust the lamina-level “properties” to fit the laminate-level test data exposed the invalidity of all such theories that relied on this sham, the process of faking the predictions this way has become institutionalised to the point that it is accepted by all involved except for the few people with a sufficient understanding of the mechanics of fibre-polymer composites to dare to question the malpractice.
These bogus theories appear in all the textbooks on the subject, are taught by Academia, promoted in short courses all around the world, and deeply embedded in all structural analysis computer codes. For 50 years after the so-called advanced composites (born-filament and carbon-fibre reinforced polymers) were developed and began to be applied, no such improved theory existed. Now that one does, the invalidity of the simplifying assumption is clear to see for all those who are willing to look. Sadly, this has just stiffened the resistance to change; there has been no acknowledgement by the advocates of the bogus theories, and their many disciples, that they are in need of replacement. Worse, the advocates of the homogenised lamina as the basis of structural analysis have been able to ensure that the enormously costly composite material data-generation programs characterised only the lamina-level properties needed for their own theories instead of the constituent-level material properties needed for more accurate theories. This lack of the properties that were really needed has greatly inhibited all attempts to improve on the analytical capability.

The impetus to develop improved theories may have been diminished by the generation, in the aerospace industry, of simple reliable methods for predicting fibre failures, even with the simplification. After all, the fibres carried virtually all the load, while well-designed laminates suffered only structurally insignificant matrix failures before the laminates failed. The problem has always been significant matrix failures that prevent badly designed laminates or structures from allowing the fibres to develop their full strength. While the empirical models for fibre failures do what they are purported to do very well, they have no capability whatever to identify laminates and structural details that will fail prematurely in the polymer matrix. This is why, 50 years later, the need to develop, and implement, reliable methods for predicting matrix failures in composites is just as great as it had been 50 years earlier.

There is an implied acceptance that all is not well with the standard method for predicting the strength of composite laminates and structures in the development process commonly called the building-block approach. The composite failure theories are first calibrated against an enormously large number of coupon tests. Their predictions are then compared to a smaller number of tests at the element and subcomponent level, and the apparent lamina-level properties and design details adjusted until they agree. The modified theories are then compared against an even smaller number of more complex test components, and the process continued. Finally, at the top of this pyramid is the real structure, which is usually too expensive to have many replicates tested. The illusion is thereby created that the theory has been validated by this process. In truth, at best, the design has thereby been validated by test because the theories were not sufficiently scientific. The testing of the real structure can be regarded as the ultimate experiment.

In the case of the Boeing 787, the development was seriously delayed by, amongst other things, a number of structural test failures that were widely reported by the news media. Every such premature failure in the composite structure occurred in the matrix, preventing the fibres from carrying the loads expected of them. Boeing had already carefully checked their design against every approved composite analysis method, including the much vaunted fracture-mechanics models (which have their own unrecognised problems with assumed homogeneity), without uncovering any of these problems. In effect, the ultimate test in the pyramid of building-block tests failed and thereby invalidated every attempt to predict matrix failures whenever the local design details were such as to cause the matrix to fail first. The structure had to be redesigned to overcome the exposed weaknesses and retested, to ensure the safety of the structure, which was validated by test, not by any theory, at least not in regard to matrix failures. (To the best of the
author’s knowledge, there were no premature failures in the fibres. The simple maximum strain model was all that was needed to reliably guard against fibre failures. There ought now to be a greater willingness to question the acceptability of the assumption of homogeneity, at least as far as matrix failures in composite laminates and structures are concerned. But, so far, the author has yet to see any sign of it.

It is, perhaps, best to start by explaining the significance of the distinction between the fibres, which carry the load, and the resin matrix, which transfers load in an out of those fibres. This paper is not concerned with situations in which the load-carrying ability of a structure is limited by the strength of the fibres—it focuses specifically on situations in which premature failures in the matrix prevent the fibres from developing their full strength. The upper diagram in Figure 1 shows a simple skin-doubler combination, in which the basic skin is locally reinforced, to increase its strength. Both the skin and the doubler can be regarded as a multitude of extremely small fibres embedded in a resin matrix. The key message from the top of Figure 1 is that all of the load in the doubler can be transmitted to or from the skin only via a single very thin layer of resin, which has less than 5 percent of the strength of the fibres. If the run-outs on the ends of the doubler are designed suitably (which process involves far more than the common but totally invalid traditional malpractice of assuming a uniform shear stress throughout the interface), it is possible for the overall behaviour to be totally fibre dominated. But, if the resin interface becomes the weak link, because the design wasn’t good enough, the structural efficiency drops appreciably; the doubler becomes ineffective, or worse. That is one example of why it is necessary to be able to accurately characterise potential matrix failures, so that one can design them out of one’s structures. Another common example is the consequences of impact damage. Impacts can break fibres on the back side of a laminate, opposite the impact. Those fibres retain their load away from the breaks; that load is transferred into unbroken fibres via the matrix, as is shown in the bottom of Figure 1. If the number of plies with broken fibres is sufficient, delaminations will be created between the broken and continuous plies. The critical question then is will those delaminations spread? The problem that has beset composite structure designers in the past is that there has been no reliable theory with which to analyse even these simple situations, despite many claims to the contrary. How does the designer know that his design is good until he either tests it or has prior experience to rely on? That is the analytical deficiency that this paper is trying to redress.
The manufacturing and use of fibre-polymer composites structures has become widespread, despite this incomplete analysis capability, which means that inferior designs are not recognised as such until they fail in service or during structural testing. The author has devoted much of his career to overcoming this largely unrecognised deficiency in the design and analysis capability. Eventually, circa 2000, 30 years after the limitations, or worse, of the so-called interactive theories became obvious to all those who would look, he was able to collaborate with a colleague, Jon Gosse, in The Boeing Company, after the merger with McDonnell Douglas, to develop comprehensive failure criteria that did properly enable laminate strengths to be predicted when their strength was governed by matrix failures, and not just by fibre failures. This mechanistic failure model is referred to as SIFT (Strain Invariant Failure Theory); see Ref. [1] for an explanation of what this theory covers and how it works. Experimental confirmation of the SIFT model is contained in Ref. [2]. SIFT uses only intrinsic properties for the discrete fibre and resin constituents that are applied at all structural levels without having to fiddle the properties to fit higher-level tests. This paper does not focus on the SIFT model; it is mentioned only to undermine any claim that it has been necessary to assume that fibre-polymer composites were homogeneous just to be able to formulate a theory capable of predicting anything more than fibre failures. Instead, the focus is on drawing attention to what is missing from the bogus theories to explain why they are inherently incapable of accomplishing what is claimed of them, in the hope that the need to improve or replace them will become undeniable. In this context, it should be noted that SIFT is a three-dimensional failure criterion, accounting for stresses and strains in all three directions, while all of the homogenised theories are two-dimensional. This inability to account for through-the-thickness stresses and strains is why fracture mechanics analyses had to be invoked to characterise matrix failures under peel and transverse shear loads. With a full three-dimensional failure criterion, this is not necessary. The SIFT model can even predict matrix-failure initiation, which fracture-mechanics analyses cannot.

In Part 1, there is a comprehensive analytical discussion of the physical phenomena associated with the mechanics of the so-called composite materials, and clear explanations of how and why all of the most highly promoted composite failure theories fail whenever the matrix fails first. And, in Part 2, this article approaches the task of solving the problem from a different perspective than those approaches tried in the past, by drawing attention to very real physical situations for which there are undeniable needs to be able to accurately predict matrix failures, which have impeded the use of fibre-polymer composites. Hopefully, if the need for such improved failure theories becomes sufficiently obvious, something will be done about terminating the use and teaching of the many composite failure theories that are inherently incapable of predicting matrix failures, but which create the illusion that they can. Not only that, it is recommended they need to be physically extracted from existing analysis codes, to remove the temptation that they could be used in the future merely because they have been used so extensively in the past.

People who were not involved in these activities might have trouble comprehending how such a situation could ever have developed. It should be acknowledged that these problems are most acute in the aerospace industry, because their laminates contain far less resin (to boost the fibre-dominated in-plane properties), and are cured at a far higher temperature and operated in far colder environments (which generates far higher internal residual thermal stresses), than are typical in fibreglass boats and commercial fibre-reinforced plastic structures. Consequently,
Matrix failures are not so prevalent in those other industries where, in addition, there is not such an intense pressure to minimise weight by squeezing out too much of the matrix. Thus, many builders and users of composite structures have remained unaware of these matrix-failure problems. So a brief history may be needed. When fibre-polymer composites were in their infancy, during the 1950′s and 1960′s, few people, if any, knew how to predict the strength of such laminates. That was understandable, since most previous structural materials really were homogeneous; analysis methods had been developed for them many decades earlier. So the early pioneers tried, properly, to account for separate fibre and matrix constituents, using what were referred to as micromechanical models because the models were so detailed as to include individual fibres. These were complicated and cumbersome – but the worst problem was the lack of a failure criterion for the matrix. One could calculate the three-dimensional stresses, but not when the matrix would fail. (The fibres could accept load only along their length, so they were easier to characterise, even though it was not recognised then that their longitudinal strength could be adversely affected by the application of transverse stresses.) The situation was complicated further by measuring the material properties on dry fibres and unreinforced resin. Because the individual fibres failed at such variable strengths, statistical predictions of laminate strength prepared this way had too much scatter. One problem was that, when embedded in a resin matrix, broken fibres were free from longitudinal load only in the immediate vicinity of the break – and all of the breaks did not line up conveniently on a single cross section, so few of the failed fibres were significant at any one time. Worse, the unreinforced matrix was inherently free from the enormous residual thermal stresses that developed in it when it tried to shrink during the cool-down after curing, but was resisted by the relatively infinitely stiffer fibres, so the measured matrix strengths were far higher than their in-situ values.

These well-intentioned pioneers were never to be given the chance to make their analysis models workable because they were soon brushed aside by another contingent of scientists, who loudly proclaimed that they had the answers and that is wasn’t necessary to account for separate fibre and matrix constituents after all. They proposed highly elegant, but grossly over-simplified, models based on the never-verified assumption that it was permissible to assume that the composite materials really were homogeneous, or at least could legitimately be analysed as such. The use of simplifying assumptions is not necessarily wrong but, in this case, no attempt was ever made to establish the consequences of these assumptions. It was simply further assumed that, if there were any, they were unimportant. As long as there were not enough true experts to recognise the implications of this assumption, it was very appealing, because of its simplicity, which is why, more than fifty years later, there is still a need for a paper like this. At that time, boron-epoxy and carbon-epoxy composites (the so-called “advanced” composites that were so much stiffer than glass fibre-reinforced plastics that they could be looked upon as possible primary structure materials) existed only as uncur ed unidirectional laminae, which were layered in different directions to create multi-directional laminates when the lay-up was cured, almost invariably under heat and pressure. So the reference properties measured were the longitudinal strength in tension and compression, the transverse strengths in tension and compression, and the in-plane shear strength – of the lamina. Of these five properties, the first two represented fibre or filament failures, while the last three were matrix failures.

The advocates of these homogenised theories split into two camps. The engineers in aircraft companies tended to focus on non-interactive models, such as the maximum-strain and much later the greatly improved truncated maximum-strain model, see Figure 2, in which the five reference strengths were treated as separate properties, with no one interacting with any of the
other four. (The fifth property, in-plane shear, is characterised on the orthogonal axis, normal to the two-dimensional failure envelopes shown.)

Fig. 2 Maximum-Strain and Truncated Maximum-Strain Non-Interactive Composite Failure Models

Such failure theories worked well for fibre-dominated laminates because the fibres were so many orders of magnitude stiffer than the matrix that the longitudinal “lamina” strengths were really fibre strengths normalised with respect to the entire lamina cross section instead of only that of the fibres. Also, the longitudinal strain in the fibres was the same as that in the lamina. There might have been much less confusion if the in-situ fibre strengths had been expressed with respect to the cumulative cross section of tows of fibres instead. Ironically, it is common practice to adjust the measured longitudinal lamina strengths to compensate for different fibre volume fractions, which amounts to the same thing but doesn’t appear to be. (The “lamina” longitudinal strength is then a fibres-per-unit-cross-section strength standardised on a single fibre volume fraction.) However, the other three properties led to predicted laminate strengths in matrix-dominated laminates that were unacceptable in comparison with most of the test data and, even worse, consistently predicted that equivalent aluminium structures would be lighter than the composite ones. Aircraft structural engineers learned to accept this, and to simply ignore predicted matrix failures, and to prepare empirical design rules to preclude in-plane matrix failures so that, provided that these rules were obeyed, the inability to accurately predict matrix failures didn’t matter – unless one was foolish enough to ignore the possibility that they might occur if the constraints on fibre patterns and lay-up sequences were violated. Some of the non-interactive failure models, such as the maximum-strain model [3], pioneered by Max Waddoups, the truncated maximum-strain model [4], developed simultaneously and independently at several US aircraft companies, and the author’s own Ten-Percent Rule, [5], [6], and [7], will continue to be used widely, because they are simple to understand and to apply – and can predict fibre failures very reliably. This is important, since the fibres carry essentially all the load. Better yet, these models cannot be fiddled to create more acceptable answers. They are used in conjunction with empirical design rules that favoured thorough interspersal of plies with fibres in different directions and limits on the orthotropy of fibre patterns and the blocking of plies with the same fibre direction. The problem for some people, who knew more about mathematics and less about the mechanics of materials, was that these empirical design rules undermined the popular myth
that it was appropriate, or even possible, to “optimise” fibre patterns to improve structural efficiency, commonly referred to as tailoring.

There are also some homogenised non-interactive composite failures models that have a substantial but incomplete capability to predict matrix failures. The best known is that of Puck [8]. They are inherently incapable of establishing the magnitude of the residual internal thermal stresses in the matrix. Also, while the fibre failure criterion, usually the maximum-strain model, is semi-universal in the sense that the effects of different fibre-volume fractions can be accounted for analytically, the three apparent matrix-failure properties at the lamina level need to be re-established experimentally for each operating temperature and fibre volume fraction combination. What separates Puck’s works from the bogus theories described below is that he actually uses the measured lamina-level matrix properties all the way through his analysis.

The other group, predominantly in Academia, sought to overcome this inability to predict matrix failures by advocating very different interactive homogenised models for composite failure theory. The most popular of these are the Tsai-Wu criterion [10] and Hashin model [11]. (Some people believe that Hashin’s model is non-interactive, because of separate equations to cover fibre and matrix failures. Unfortunately, they are coupled by the in-plane shear stress terms, and each equation involves multiple failure modes, making them interactive.) These are the most prominent of such theories, but there are dozens of basically similar clones suffering from the same inherent deficiencies. One major problem with trying to predict matrix failures by analyses at the homogenised lamina level is that the transverse strains, for example, in the matrix are not the same as those in the lamina. Worse, it is assumed that there are no through-the-thickness stresses unless external loads are applied in that direction. Worse still, it is assumed that there are never any longitudinal matrix stresses of significance, even though those actually cause the greatest thermally induced residual stresses, because the fibres resist contraction of the matrix the most along their length.

The deficiencies of the interactive theories are aggravated because the matrix is riddled with substantial thermally induced residual stresses – none of which could occur in a truly homogeneous material. These are caused by the intense shrinkage in the resin matrix as it cools down from it very high cure temperature to the sub-zero temperatures at high altitude and in space, with the contractions resisted by the very much stiffer fibres that effectively do not shrink at all. There are absolutely no terms in the interactive theories to account for this major reduction in the capability of the matrix to withstand mechanically applied loads.

Even the title of Ref. [10] refers to anisotropic materials, not fibre-polymer composites, but this has in no way inhibited its misapplication to the latter. Indeed, despite the disclaimer implied by the title of the paper, the reason for preparing the paper at that time was to use the theory to analyse fibre-polymer laminates. The Tsai-Wu model may very well be a suitable characterisation of what really are homogeneous anisotropic solids, like rolled metallic plates and extrusions. In those cases, there is only one material involved and a common “failure” mechanism, yielding (distortion), all the way around the failure envelope, as in Hill’s well known plasticity theory. Nevertheless, the Tsai-Wu model has zero relevance to fibre-polymer composites because two materials are involved and four different failure mechanisms. No single equation can characterise that situation. Hashin had recognised the fatal flaws in interactive homogenised failure theories as soon as Tsai started to promote them. That is why Hashin offered his own theory [11], not recognising that it, too, was interactive because his two

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1 An English-language presentation of Puck’s work can be found in Ref. [9].
equations were coupled by the in-plane shear stresses. Many years later, Hashin finally understood that all such theories, including his own, were invalid. He then switched his research activities to a totally different field. This is the reason he gave for not participating in the World-Wide Failure Exercise to compare the various composite failure theories, Ref.’s [9]. His explanation of why he declined to submit predictions for his theory is included in one of the papers published by the organisers. His reasons are cited on pp. 12-13 of the book containing all of the papers in the exercise. The key sentences are “I do not believe that even the most complete information about failure of single plies is sufficient to predict the failure of a laminate, consisting of such plies,” and “I must say to you that I personally do not know how to predict the failure of a laminate (and furthermore, that I do not believe that anybody else does)” 2. Surely the omission of such a prominent composite failure theory as Hashin’s from the WWFE should have raised some questions. Yet his theory continues to be used widely and is embedded deeply in computer codes. It is also the most common argument presented to the author, by the members of the Composites Establishment, that there is no need for the SIFT model, or for any other constituent-level model, because existing theories, which the author regards as bogus, are apparently already able to predict matrix failures. The original intent of the WWFE was to have the authors of many of the more prominent composite failure theories use their own theories in a blind test to solve the same problems using the same material properties. This way, there could be no doubt that the theories had been applied correctly and that the predictions were comparable. Also, there was never any intent, initially, to allow for a revised analysis once the test data had been revealed, before the theories were compared. (The Maximum-Strain Theory was an exception; its originator had transferred to a totally unrelated occupation and was unable to participate in the study. The author volunteered to include that theory, as well as his own, since the Maximum Strain Theory was too important to have been omitted from the exercise.)

Because the process of comparing theories with experiments, in the context of fibre-polymer composites, has been so thoroughly discredited by the advocates of interactive composite failure theories by “modifying” the measured lamina reference properties into totally different parameters that cannot be related to any measured properties in order to show “agreement” with laminate test results, it is necessary to rely extensively on simple analogies that cannot be misinterpreted to explain what is really going on. This is why composite failure theories are explained later in the context of steel-reinforced concrete structures. A simple analogy with which to understand the meaning of the word homogeneity is as follows. Individual constituents of blue and yellow paint can be mixed to create a homogeneous “composite” in the form of green paint. Yet, if oil and water are mixed together, the separate nature of the two constituents remains visible. In this case, there is no stable homogenised composite of the two ingredients. When fibres are embedded in a resin matrix, the separate constituents are clearly visible in any photomicrograph of the cross sections. This is clearly not a homogeneous composite, either, despite being called such.

Mathematically elegant interactive models were such as that shown in Figure 3, in which the lamina strength was characterised by a single continuous curve passed through the four unrelated in-plane lamina reference strengths. This is characteristic of all of the composite failure theories that the author labels here as bogus. They all share two characteristics; the fibres and resin

2 The SIFT failure model was not yet developed. It could not be included in the WWFE exercise at the time without first determining experimentally the necessary critical values of the two strain invariants for the specific resins used in the test data supplied. The critical invariant values were not needed for any of the theories actually involved. Sadly, these properties are still not available and, consequently, the SIFT failure model has yet to be assessed as part of this exercise.
constituents are homogenised into an allegedly equivalent material, and the failure prediction involves an inter-relation between totally unrelated failure mechanisms. This definition therefore excludes all the homogenised non-interactive models, mentioned earlier, whether they account for only fibre failures or for matrix failures as well.

Now, since it requires five reference points to define an ellipse, not four, an infinite number of ellipses can be passed through those four in-plane reference strengths. (The fifth measured value, in-plane shear strength is associated with a third axis, perpendicular to the plane shown in Figure 3.) Therefore, some assumption must be made to define the in-plane portion of the ellipsoidal failure surface. This, of course, cannot be tied to any measurable lamina reference strength, but can be established by a curve fit through a series of biaxial measured reference strengths. The inherent weakness of this approach is that the position of every point off the axes is affected by all four of the reference strengths. Curves such as that in Figure 3 are nothing more than meaningless curves passed through unrelated data points, as is clarified by the scientifically precise analogy shown in Figure 4.
The *fatal* flaw in the approach in Figure 3 is that it is not possible to identify, around this curve, any points at which the failure mechanism changes from one in the fibres to a different one in the matrix. In other words, there is no way of knowing *which* constituent is failing. It is important to know that since, when the fibres fail first, there is no residual strength in that ply. But, when the matrix fails first, locally rather than globally, the fibres can sometimes continue to carry load. This, of course, is significant only in the context of multi-directional laminates. It rapidly became obvious that low predicted laminate strengths could be enhanced by artificially reducing the mathematical transverse lamina modulus, which is clearly a matrix-dominated lamina property. This effectively suppressed the prediction of matrix-dominated failures, allowing the stronger fibre-dominated failures strengths to be predicted instead. It was important not to reduce the transverse ply stiffness all the way to zero because, then, the theory would be transformed into the maximum-strain model and it was vital in marketing the interactive theories to show some differentiation with respect to simpler empirical methods. Much the same effect could be achieved by increasing the transverse strain to failure. These arbitrary modifications of the transverse properties could be “justified,” or so it seemed, if non-catastrophic matrix damage could be equated to a reduction in transverse lamina stiffness. And so was born the concept of progressive-failure analyses, to restore credibility to the interactive composite failure models. This subterfuge would transform the unacceptably low original predictions, based on the measured lamina reference strengths, referred to as first-ply failures, into much higher last-ply failures by arbitrarily changing the transverse lamina properties used in the analyses. To reinforce the deception it is assumed that all of these early matrix failures within individual plies will have no effect on the ultimate laminate strength, when the fibres finally fail. They are not accounted for when strength margins are written, for example. Such matrix damage is not perceived as being real, which is why it is “permissible” to change the lamina properties so that matrix failures are not apparent in the model used to predict fibre failures in the laminate. To the best of the author’s knowledge, no one has ever attempted to actually measure the reduced transverse modulus of the laminate in which certain plies had been predicted to fail. Instead, the reduction in modulus has been deduced on the basis of how much was needed to produce an acceptably high predicted last-ply-failure laminate strength. Indeed, in one such theory, the reduction in modulus was contained in an iterative loop, with a hundred-fold reduction in modulus per pass, until the last-ply failure prediction stopped increasing, by which time the original theory had been transformed into the maximum-strain model with matrix-failure predictions suppressed, without acknowledging that this is what had been done.

The problem of differentiating between fibre and matrix failures in the homogeneous material with no separate constituents was solved by simply *defining* parts of the failure envelope to be fibre failures, with the rest defined to fail in the matrix. Only the portions defined as matrix failures, usually the entire transverse tension domain, are eligible for re-analysis with modified properties. There is no experimental confirmation for this, either; indeed, there is one obvious contradiction. The tests show that the matrix fails under pure transverse compression of the lamina, even though this technique specifies that the fibres must fail first, even though there is then no longitudinal load in them. The hypocrisy of this technique is exposed every time the last-ply failure is predicted to *precede* the first-ply failure, as described in Ref.’s [12] to [15]. Nevertheless, these mathematically elegant interactive composite failure theories are still promoted strongly by much of Academia and embedded deeply in all the text books, computer codes and short courses used for structural analysis. The vested interests in perpetuating such nonsense have, so far, triumphed over every attempt to change to more realistic and reliable failure models that do *not* need to be fiddled to generate an acceptable prediction. There is a
simple challenge to pose to the advocates of such highly questionable analysis techniques; “How does one really differentiate between fibre and matrix failures inside a homogeneous anisotropic solid that cannot possibly contain either fibres or resin matrix because, if it did, by definition, it could then not possibly be homogeneous?”

The form that Figure 3 SHOULD have had is shown in Figure 5, based on the failure envelope associated with the SIFT model, which is non-interactive and completely defines which constituent fails where. This truncated envelope, at the lamina level, is based on the intrinsic in-situ properties of the fibres and resin. It is not the result of fiddling lamina-level properties to suit. It is important to note that each failure mechanism is fully defined by the single data point associated with that mechanism, as the laws of physics would require. There is no fitting of arbitrary curves through multiple data points. These two characteristics do not change much with fibre volume fraction or operating temperature – but the location of the cut-off in Figure 5 does!

Some might argue that Figure 5 proves that matrix failures can be characterised successfully at the lamina level, and that this undermines the author’s criticisms of the bogus theories. It is true that one can combine the maximum-strain model for the fibre with a transverse-tension stress cut-off for the matrix. But this is a non-interactive model, and therefore has no bearing on the interactive theories the author has condemned. Also, while the fibre failure can be universal, the transverse tension matrix cut-off isn’t. It varies with fibre volume fraction and operating temperature, so, without SIFT, this cut-off would need to be re-established by test for every circumstance.

It will not suffice to simply add new composite failure theories that actually do obey the laws of physics to the seemingly infinite number of existing theories that contradict them. Not even
adding warnings about the danger of using such theories, as are now being mandated for cigarette packs, will convey the importance of the need to stop using the bogus theories. And they would probably be equally ineffective. The latter theories are so widely applied as a simple black box – garbage in, gospel out – that progress in the application of composite materials will continue to be impeded until all these interactive failure theories are relegated to academic ideas that looked good but didn’t work AND all of their associated computer codes are physically deleted from both newly sold structural analysis codes and all those already installed on computers worldwide. Otherwise, new composite products will continue to be plagued by the reliance on expensive and time-consuming tests for certification. And, even those cannot cover all environmental effects fully. Every such problem overlooked at the design stage will cause delays in introduction to service as they are solved and fixes are applied. The cost of the lack of reliable methods for predicting matrix failures in composite structures is no trivial matter. Not only is there a high economic cost, the absence of universally accepted workable theories with which to quantify laminate and structural strengths at which matrix failures occur has led to the adoption of structurally inferior fibre patterns and ply lay-ups, and poor design details – and a widespread lack of awareness of what needs to be changed to improve the laminates.

The summary above explains in physical terms what is wrong with the status quo for analysing matrix failures and designing fibre-polymer composite structures. This information has been widely available for well over 30 years, and the SIFT model was developed over 10 years ago. This has not been sufficient for something to be done about the problem – or at least no more quickly than the time it took for the Roman Catholic Church to acknowledge that the Earth really wasn’t flat! The parallels with religious beliefs run even deeper than this. The advocates of the bogus theories expect their disciples to accept that fibre-polymer composites are homogeneous entirely on faith, because there is no way of substantiating this claim scientifically. It is not even admitted that the assumption of homogeneous laminae is a simplifying assumption, or that it has never been validated.

Part 2 of this document focuses on a different approach to accelerating change – by explaining the serious structural problems that arise because of the continued acceptance of composite failure theories that are not only inherently incapable of predicting matrix failures, but at times capable of making dangerously misleading estimates, with no warning to not believe them. The hope is that frustrated customers who actually used the bogus composite failure theories will exert enough pressure to overcome the past resistance against improvements, which the author has repeatedly experienced first hand.

THE INSUPERABLE PROBLEM OF PREDICTING PATH-DEPENDENT STRENGTHS AS A CONSEQUENCE OF “PROGRESSIVE DAMAGE” ANALYSES

At this point, it is necessary to make what might appear to be a digression from the main theme of the paper. However, it really isn’t. It is an issue of critical importance in the design process, related to matrix failures, that has been steadfastly shunned by almost everyone involved in composites ever since the author first pointed it out some 25 years ago – because there is no possible answer to the issue raised that is acceptable to an advocate of composite structures as being inherently superior to metals, always, rather than just another engineering material that is suitable for some applications and unsuitable for others. In a nutshell, when real subcritical matrix failures are tolerated, or phony ones concocted, the final predicted laminate strength, when everything fails, occurs at a load that varies with the prior history of previous non-catastrophic failure loads. No designer could cope with a material having such ambiguous
properties (strengths and stiffnesses), yet this is exactly the situation created by the advocates of progressive-failure theories, in which the first one or more “failure” loads are not catastrophic. This approach is what is meant by first-ply failures and last-ply failures, which are the lynch pin in salvaging the interactive composite failure theories criticised above.

Progressive-failure analysis is used roughly as follows. A load is applied until something is predicted to break, using the measured lamina properties for the analysis. If the failure is not catastrophic, the internal stresses and strains are assessed and those laminae in which the matrix has been predicted to fail have their properties modified (zeroed-out in some procedures). The nature of the modification is irrelevant to the present discussion; all that matters is that some ply properties have been changed. A second load is applied, using the measured properties for the undamaged plies and the modified ones for those with matrix damage. When something else is predicted to fail, the same procedure is repeated – until what fails are the primary load-carrying fibres, at which point there is no residual strength. It is not necessary that the subsequent loads be the same as the first one, only larger. Indeed, for a wing structures, subjected to up-bending, down-bending, and torsion in a random sequence and with random magnitudes, it would be necessary to cover dissimilar sequential loads. Even worse, matrix failures can result from impact damage which, by its nature, is random in nature, intensity, and frequency of occurrence. The problem with this scenario is that the magnitude of every subsequent load when some failure is predicted to occur is then a function of the ply properties (assumed or real) which are, in turn, a function of the location, through-the-thickness and within the plane, of any matrix damage that may be present. This, in turn, is a function of the sequence and nature of the previously applied subcritical loads. In other words, the next failure load is effectively undefined!

The only exception to this dilemma that the author has been able to identify is situations for which there is only one load, as in a pressure vessel. In these cases, if one could predict matrix failures reliably – and knew how to characterise the appropriate change in ply properties as a result of that matrix damage – one could legitimately take credit for the increased residual strength (provided that one inserted an impermeable internal bladder to stop the contents from escaping) – and be able to rely on the estimated ultimate strength. That is common design practice in this context, based on sound empiricism, not a rigorous theory.

Otherwise, there seems to be no way to apply progressive-damage theory in a rational manner. Maybe it will now be apparent to readers who could never understand why the U.S. Navy would not tolerate matrix failures – and defined first-ply predicted failure to be design ultimate strength – why they felt that way. Somebody at Warminster must have reached the same conclusion that the author did.

As a minimum, the discussion above should make it unambiguously clear how important it is to be able to accurately predict both the occurrence of matrix damage and its effect on the residual strength of composite laminates. Even the traditional assumption of the consequences of non-catastrophic matrix failures within some plies is contrary to reality. When that damaged ply is embedded in a multidirectional laminate, the loss of transverse stiffness is restricted to a microscopic strip containing each individual crack. The remainder of the ply, the great majority of it, retains its full stiffnesses in both directions, so the transverse lamina-level strains to failure would be unaffected. The effect of these cracks would therefore be undetectable at the laminate level, as distinct from that ply extracted from the laminate and loaded in isolation, when the “first-ply” failure would then be catastrophic, i.e. it would then become a “last-ply failure.” So there would be no possible justification for modifying the transverse ply properties for the next
analysis, which is based on stiffnesses to distribute the internal loads. This would then predict exactly the same thing that the previous analysis had. But, acknowledging this would totally undermine the subterfuge of progressive failure, or ply discounting, analyses to “justify” faking the transverse ply properties to generate a more acceptable answer.

**EXPLANATION OF THE DIFFERENCE BETWEEN HETEROGENEITY AND HOMOGENEITY VIA AN ANALOGY WITH REINFORCED CONCRETE**

Before explaining how residual stresses develop inside the matrix of fibre-reinforced plastics, it is necessary to make unambiguous the difference between artificially homogenised truly heterogeneous composites of materials and other more conventional materials that actually are homogeneous. In retrospect, it will appear that no one could be so naive as to have ever believed that there was no difference; it is almost as if there were a conscious belief that the laws of physics did not apply to fibre-polymer composites because carbon fibres were created by man, rather than being a product of Mother Nature. The author apologises in advance to those readers who do not need this explained to them, but asks that they do accept that there are other people, including a lot of acclaimed experts on composites, who obviously do.

The need to be able to design and analyse structures on the basis of the constituents that make up the material being used are easily seen by examining a slab or concrete, with steel reinforcement at the top and bottom, and trying to analyse it using the rules established for homogenised fibre-polymer composites. The model is shown in Figure 6.

![Fig. 6 Physical Model of Unit Cell of a Steel-Reinforced Concrete Slab](image)

Assume that the overall stiffnesses and strengths of this piece of slab can be measured, on the assumption that it can be replaced mathematically by an equivalent homogeneous block, with different moduli and strengths in each direction. There would be a high longitudinal Young’s modulus, in the direction of the reinforcing rods, a low transverse modulus in the direction perpendicular to the rods, a low in-plane and transverse shear modulus, a high longitudinal strength in tension and compression, a low transverse strength in tension, a not-quite-so-low transverse compression strength, and a low in-plane and transverse shear strength, just as there are for unidirectional fibre-polymer composite laminae. Given only these (never measured)
properties, one could calculate the average stresses associated with any given strain level – and one would not need the individual properties of steel or concrete.

But, now, try to calculate the bending stiffness of the slab using only those homogenised properties. One can’t! One needs the individual constituent properties of steel and concrete. The process is then quite straightforward. The concrete slab is replaced with a series of layers, as shown in Figure 7, with the individual steel rods replaced by a simpler single sheet of steel at the appropriate depths, and with the same total cross section. This is a realistic model for slabs, which are reinforced in each direction. One would ignore the added transverse strength and stiffness of these layers for a beam, because of the absence of rods in that direction, so the Young’s modulus of the steel in that direction would be set to zero for beams, and to its normal value for plates, which are reinforced in both directions. The in-plane shear modulus of the steel layers would also normally be set equal to zero, so that all of that load would conservatively be assigned to the concrete.

Fig. 7 Mathematical Model of Layered Unit Cell of a Steel-Reinforced Concrete Slab

The membrane and bending properties of this equivalent layered block can now be calculated by conventional rules of mixtures to the individual properties of steel and concrete, accounting for their respective volume fractions and their locations through the thickness. This approach is universally accepted for designing and analysing reinforced concrete structures. The theoretical shear strengths may be a bit low if the steel is discounted for all but longitudinal loads, but not by much since the reinforcing volume fraction is small. The same approach is even adopted for what is called lamination theory for laminates composed of a series of layers with fibres in different directions, except that the fibres and resin within every layer are artificially homogenised.

If the steel and concrete were not segregated into individual layers with very different properties in Figure 7, but smeared into an “equivalent” homogenised single block, as is done for the fibres and resin in so-called composite materials, the in-plane properties for a beam would appear as something like the failure envelope shown in Figure 8. This is obviously absurd for steel-reinforced concrete. But why is this not equally obvious for fibres embedded in a resin matrix?
The longitudinal tensile and compressive strengths of this artificial homogenised lamina are so dominated by the steel reinforcing rods that they are independent of any transverse loads the concrete would withstand, so the “failure envelope” in Figure 8 should really have been a box, with vertical ends, as in the maximum-strain model for composites.

What is important to note about the actual methods used to design and analyse steel-reinforced concrete is that the mathematical model did NOT require that each individual steel rod be modelled separately. This gives the lie to the claim, by the protagonists of using the homogenised-equivalent-anisotropic material for the design and analysis of fibre-polymer composites, that the only possible options were complete homogenisation or the inclusion of every single fibre, with no possibilities for alternatives anywhere in between. There were always intermediate possibilities.

Why is it so obvious that the concept of a homogenized “equivalent” steel-reinforced concrete model makes no sense while it is insisted that exactly the same model is appropriate for fibre-reinforced resin composites?

How does encasing the steel rods in concrete increase their longitudinal compressive strength when subjected to transverse compression?

**THE MISCHARACTERISATION OF EDGE DELAMINATIONS AT PLY INTERFACES IN FIBRE-POLYMER COMPOSITES**

Perhaps the most striking example of this inflexible attitude is to be found in the analysis of edge delamination stresses in fibre-polymer composite laminates. The earliest analysis that the author has found was a little-known paper [16] in which the composite laminate was replaced, mathematically, by a series of layers, as shown in Figure 9.

Each layer of resin-impregnated fibres was replaced by an equivalent homogenised lamina, separated by a very thin, but greater than infinitesimal, layer of resin. This model predicted that there would be finite peak stresses in the resin layers each time there was a change in fibre direction between adjacent plies. Their paper had scarcely been published before it was followed by what is regarded as the definitive pioneering reference of the subject [17], with the earlier model modified by the omission of all of the resin layers, as shown in Figure 10.
Not surprisingly, this over-simplified, but now almost universally accepted, model predicted *singularities* in the peak stresses at the zero-thickness interfaces between the plies, at the edges of the panel, whenever the direction of the fibres changed. To indicate how devoutly the advocates of homogenisation believe in their approach, the highly-acclaimed experts who wrote the second paper had the unmitigated gall to criticise the lesser mortals who had written the first paper because of their inability to predict the singularities! Whole careers have been devoted to characterising these fictitious singularities. All of the innumerable fracture mechanics-analyses of this problem break down if the thin interfacial resin layers are restored to the models. This is just one of the problems besetting the predictions of matrix failures in fibre-polymer composites, but it is indicative of the basic fundamental errors caused by the pretence that fibre-polymer composites either are, or may legitimately be considered to be, homogeneous anisotropic solids. Another inevitable consequence is that the homogenised models predict that higher fibre-dominated strengths can be created by squeezing out even more of the resin matrix – which causes even earlier failures in the matrix, which go uncorrected because the theories used are incapable of reliably predicting the degraded matrix-dominated strengths.

*Fig. 9 Contrarian Model of Layered Unit Cell of Fibre-Polymer Composite Laminate With Interfacial Layers of Resin*
EXPLANATION OF THE ORIGIN OF RESIDUAL THERMAL STRESSES IN THE RESIN MATRIX OF FIBRE-REINFORCED PLASTICS

The origin of residual thermal (and sometimes also chemical shrinkage) stresses in the matrix for fibre-reinforced polymers is easy to understand once one is willing to accept that fibre-polymer composites are not homogeneous, but contain two discrete constituents with very different properties. In particular, they have very different thermal and mechanical characteristics. The resins have a very high thermal coefficient of thermal expansion, while carbon fibres have an almost zero coefficient along their length and a small value in the transverse direction. (These fibres, unlike glass, which is isotropic, really are homogeneous anisotropic materials.) The resins have a modulus that is extremely low in comparison with the longitudinal modulus of all the traditional reinforcing fibres, while the transverse modulus of carbon fibres is very small in comparison with the longitudinal value, but significantly higher than that of the typical resins.

The generation of the residual stresses in the matrix surrounding the individual fibres is most easily comprehended in terms of the chemical shrinkage associated with the transformation of the resin from a liquid to a solid. In some polymers, this can be as much as a few percent, which is why they are not used for these aircraft structures applications. But even the resins that are used have typically one percent or so chemical shrinkage. In the absence of fibrous reinforcement, a unit cube of resin would shrink unrestrained, during the cure, by this small amount – equally in all three directions. However, when very stiff fibres are embedded in the resin, this shrinkage is resisted by the fibres in all three directions and virtually totally suppressed along the length of the fibres. This resistance creates stresses, and strains, in the resin because, in effect, the freely shrunk resin must be enlarged to ensure compatibility with the fibres. This is explained in Figure 11.
These residual stresses obviously detract from the mechanical strength of an unreinforced block of resin. Yet, using the homogenised model for composite laminae, no such stresses can exist. So, any prediction of matrix failures based on this model must inevitably be wrong! Quantifying just how wrong cannot be established any way other than by using a fibre-polymer model containing discrete fibre and resin constituents. Sadly, there has been a dearth of models with which to quantify the errors associated with the over-simplified homogenised models, which therefore remain largely unseen.

There are also appreciable residual stresses induced in the matrix as it cools down to operating temperatures (room temperature and far below that at high altitude) after being cured at high temperature. Apart from the chemical shrinkage stresses, the constituents of the laminate are essentially stress-free at some temperature close to when the polymer cross-links and solidifies. Then, during cool-down, the solidified resin tries to shrink, but is resisted by the embedded stiff fibres which have very much lower coefficients of thermal expansion than the resin. In the case of carbon fibres, there is essentially zero thermal expansion or contraction along the length. The net result is a state of high residual thermal stresses in the laminate, as described in Figure 12, with very high tensile stresses along the length of the fibres, appreciable tensile hoop stresses around the fibres, significant radial compressive stresses normal to the fibres, as well as significant tensile normal and tangential stresses (on top of the ever-present high longitudinal stresses) throughout the interstices between the fibres. These triaxial thermal stress increments obviously cannot be accommodated in a model of a two-dimensional homogenised layer.

Strictly, the criticality of the condition in the matrix should be expressed in terms of strains, specifically the first two strain invariants – those of dilatation (increase in volume) and distortion (shear). However, Figure 12 relies on stresses to identify these regions, since so many prior failure theories have been expressed in terms of stresses, and because more engineers seem to understand stresses than strains. Figure 12 also identifies the regions that are most critical under transverse mechanical loads, to emphasise the point that the most critical locations for mechanical and thermal loads are not always the same.
These internal residual stresses in the matrix can be quantified by the SIFT failure model [1] and [2]. The reason why the homogenised lamina models are inherently incapable of doing so is that matrix failures within a lamina or laminate are governed by what is left of the matrix strength after the residual stresses (or strains) associated with chemical shrinkage and thermal contraction have been subtracted from the inherent capability of the unreinforced resin. The claim by some of the proponents of homogenised composite failure theories that they can be ignored because they creep out with time can be dismissed as mere arm waving, since it should be obvious that this is a slow process, so internal thermal stresses would always be created every time there was a relatively rapid change in operating temperature. Worse, if the creep really did occur as “required”, it would also occur on long trans-oceanic flights, leaving very distorted structures after landing. The other argument offered, that the residual stresses are accounted for in the measured properties, is specious, since the matrix strains associated with thermal and mechanical loads interact nonlinearly in regard to distortional failures. (The dilatational contributions combine linearly, but only when the analysis is formulated in terms of the three orthogonal strain components, not in terms of stresses.)

All of these chemical shrinkage and thermal stresses are present in the matrix whenever lamina strengths are measured. But there are no terms in the homogenised theories with which to account for them. Obviously the thermal stresses are intensified in aircraft structures flying at high altitude, where it is extremely cold. These residual thermal stresses cannot possibly be measured directly – and they cannot be “compensated for” by assigning slightly lower fibre-dominated reference strengths, and testing only at room temperature, as is common practice because of the enormous cost of performing structural testing on complete structures at the real sub-zero operating temperatures on complete structures or even subcomponents large enough to allow real matrix failures to occur.
The fundamental problem with accounting for real matrix failures in fibre-polymer composite laminates and structures using analyses based on the false assumption of homogenisation is that the variation in lamina strength with temperature is totally different from the variation in residual mechanical strength of the matrix, after the chemical and residual thermal stresses have been subtracted. Unidirectional lamina strengths are usually measured to be about 10 percent lower at -55°C, (-67°F) than at room temperature, (although this is actually largely due to the difficulties of testing brittle materials, since the fibre strengths do not exhibit that change), while the remaining mechanical strength of the matrix has already been reduced by a factor like 1.5 between the stress-free and room temperatures and a further factor almost as high for operation at high altitude. The notion that the residual matrix stresses can always be ignored when predicting composite laminate strengths by using over-simplified homogenised material models is patently absurd.

INFLUENCES OF SIZE EFFECTS ON MATRIX FAILURES IN FIBRE-REINFORCED PLASTICS

The mechanical properties of a unit block of metal (or any other truly homogeneous material) are the same as those of a block ten times as large. However, not all the properties of a unit block of fibre-reinforced plastic are the same as those for a scaled up block containing the same number of layers – strictly, interfaces between changes in fibre direction. But the homogenised composite failure theories would predict that there would be no difference. (Fracture mechanics analyses could “predict” such a difference, but only after new test measurements were made for each ply thickness – and for each different operating temperature. They are not capable of being generalised as they would have been if they were dealing with truly intrinsic material properties.) This so-called size effect is manifest in many forms – all associated with matrix failures that would prevent the fibres from developing the same high strengths for all tow sizes and ply thicknesses. This has led to unanticipated premature failures, because of the myopic faith that part-count reduction will always solve all economic problems with manufacturing high-technology composite products – and almost everything else! The absence of accepted theories with which to demonstrate the folly of such thinking has led to the perpetuation of such problems, which will continue unabated until rational cost-estimation procedures that take account of the laws of physics are established. This blind focus on part-count reduction has led to innumerable co-cured composite structures that have consistently been found within McDonnell Douglas to be more expensive than equivalent (but superior) structures made from the secondary bonding of multiple simple components in place of each over-complicated one-shot structure, by factors of between 2.4 and 3 whenever there has been a one-to-one comparison available to reveal actual costs. Ironically, many of the one-shot cure designs are so complex that it has required up to five sequential steps, and five visits to the autoclave or oven, to even make a part of sufficient quality that it didn’t need to be scrapped. (It could never have taken more than two steps to make simple details in a first cure and bond them together in a second step, using tooling some orders of magnitudes simpler and less expensive than those needed to locate uncured stiffening details). This dogma is so deeply entrenched that the author believes the powers-that-be would try to machine entire Boeing 747 fuselages from a single billet of aluminium if they knew how – because their cost-estimation processes would predict that doing so would reduce costs in comparison with assembling the fuselage from pieces. As a humorous anecdote to confirm how deeply ingrained this attitude is, there once was a study to replace some floor beams assembled from many sheet-metal and extruded parts by an integrally stiffened one-piece machined part, even though the billet from which the one-piece part was to be machined cost
more than the finished built-up beam, even before the cost of the energy to convert more than 95 percent of the slab into chips was accounted for. One needs to understand this background to comprehend efforts to speed-up the manufacture of composite aircraft structures by laying down composite material in a smaller number of larger slabs.

The physics behind these problems are explained in Figure 13. The key parameter is the ratio of fibre (or tow) cross section to the perimeter through which the resin can transfer load in or out of that fibre (or bundle of fibres). This ratio is inversely proportional to the size of the fibre (or bundle of fibres), so it is infinitesimal for individual fibres but increases steadily with the number of adjacent parallel fibres. Eventually, the cross section becomes so large that the inherently weak resin matrix is incapable of transferring the load through the interface, so the fibres can no longer be loaded up and the interfaces will fail (split parallel to the fibres) around the edges. This situation can be characterised only by acknowledging that the fibres and matrix are discrete constituents of the composite of materials. This mathematical problem simply cannot exist when the “matrix” has the same properties as the “fibres” because, then, there are no interfaces.

**Fig. 13 Explanation of Size Effect (Tow Size) in Transferring Interfacial Shear Loads Between the Matrix and the Embedded Fibres**

The absence of any theory with which to discourage the use of larger tow sizes, and ply thicknesses – or even worse, the blocking together of parallel plies instead of interspersing them – has led to a drive to inadvertently ensure that the “lower-cost” laminates and structures would be far weaker than higher-cost laminates in which the matrix was given a better chance of allowing the fibres to function efficiently. Ironically, when laminate strengths are restricted by premature matrix failures, there is an apparent need to add *more* equally thick plies of material, increasing both the cost and weight of the structure. The option of dispersing the existing fibres better passes unrecognised because there has not been any model that could be encoded to create a computer print-out quantifying the benefits of the superior design. (This particular problem is far more pervasive. For example, the merits of increasing the overlap in longitudinal lap splices in aircraft fuselages went unnoticed for over 30 years because the standard linear analysis showed no sensitivity to the overlap length. It required a geometrically nonlinear analysis to demonstrate the need to do so – and the improvements thereby achieved. The situation is
identical for matrix failures in fibre-polymer composites; they don’t “exist” unless the computer print-out says that they do. But they happen, nevertheless! )

THE SO-CALLED KEVLAR/NOMEX PROBLEM OF MOISTURE ABSORPTION AND ACCUMULATION IN THE CELLS OF HONEYCOMB FACE SHEETS

The widespread failure to understand this situation is well demonstrated by the misdirected efforts to solve the so-called Kevlar/Nomex problem affecting whole fleets of aircraft containing secondary structural fairings made from these materials about 30 years ago. The panels tended to fill up with absorbed water during successive changes in altitude from the ground to cruising flight. Eventually there would be so much water accumulated that the water would freeze at high altitude and expand to delaminate the panels. While not a safety issue, it was an economic disaster because of so many planes having to be taken out of service for repairs. The expert committee solving the problem totally missed the dominant factor. They believed that the root cause of the problem was the use of a brittle primer that was cracking on the surface, allowing moisture to slowly migrate through the underlying resin and to collect in the cells of the honeycomb core. The dominant factor was actually the use of Kevlar filaments in coarse-weave thick layers, with large tow sizes to “reduce cost.” The first “cracks” that allowed the moisture to collect rapidly in the cells were actually in the essentially dry fibre tows inside the laminate, not on the surface, as explained in Figure 14. Kevlar fibres are hard to wet under the best of circumstances, and there is no way for the resin to penetrate into the interior of the large bundles of fibres in each large tow. So, once the moisture had penetrated through the outer surface, there was a wide-open path straight through to very close to the inner surface of each facing. Significantly, the cure to this problem was to coat the panels with a thin layer of fine-weave layer of fibreglass cloth when it was being laid up. This was essentially crack-free and provided an adequate moisture barrier.

![Diagram of moisture accumulation](image)

Fig. 14 Phenomena Related to Water Accumulation in Kevlar®/Nomex® Honeycomb Sandwich Panels Via Cracks through the Face Sheets
The author believes that the reason why the investigating team missed the major factor was that they weren’t thinking of the fibre-reinforced plastic face sheets are separate fibres and resin, but as homogenised, and therefore initially crack-free, layers. This was particularly surprising, since the surface cracks lined up perfectly with the underlying tows of fibres, with a ladder-like distribution of cracks on the surface. It should have been obvious that they had initiated from within the laminate and propagated to the surface. The author was led to his understanding of what was happening by fatigue tests on carbon-epoxy test coupons with woven $0^\circ/90^\circ$ plies at high stress levels. One set of such coupons was coated with a light grey paint and the small surface cracks showed up as fine black lines arranged exactly as they had appeared on the Kevlar/Nomex panels. But the test coupons did not have a brittle primer, so it could not have been a factor. These cracks stopped, changed direction, and started again at each change in direction of the fibres closest to the surface, matching the weave of the fibres perfectly.

**CONCLUDING REMARKS**

The author hopes that this paper, with Part 2 illustrating typical situations of matrix failures preceding fibre failures, has presented a convincing case that it is necessary to be able to predict matrix failures in the so-called composite materials reliably – and, further, that there is no possibility of doing so with theories that have been oversimplified by the never-verified assumption that it was permissible to homogenise the discrete fibre and resin constituents into an “equivalent” anisotropic solid. It is necessary, whenever the matrix fails first, to characterise the distinct fibre and polymer constituents separately in the analysis models.

This simplification of homogenisation is valid for calculations regarding the elastic constants of laminae and laminates, and has permitted the development and use of valuable simple empirical non-interactive analysis models for predicting fibre failures in such laminates. No criticism is implied here in regard to these non-interactive theories which, for well designed structural details, are all that is necessary to complete the designs. What makes this possible is the enormous disparity between the strengths and stiffnesses of the fibre and matrix constituents whereby, in well-designed composite structures, almost all of the load can be carried by longitudinal stresses in the fibres. Under these circumstances, the strains in the fibres are reasonably well approximated by those in the laminae, with the longitudinal strain being dominant. The reason why this does not apply for matrix failures, which occur almost entirely as the result of poor design practice or damage in service or during manufacture, is that the strains in the matrix are very triaxial in nature and not at all well approximated by those in the laminae. Worse, much of the inherent strength of the resin matrix is consumed by intense residual thermal stresses – which simply cannot exist in truly homogeneous materials, so their presence is totally concealed by using any theories that do not account for the discrete fibre and resin constituents of real composites of fibres and polymers. In other words, it is impossible to characterise the state of stress, or strain, in the matrix using the popular artificially homogenised representations. That is the one limitation of the non-interactive composite failure models; while they allow adequate analysis of fibre-dominated structures, they give no warning that other weaker structures might be matrix dominated.

The author acknowledges that he is not alone in trying to improve the capabilities of composites analysis models that can account scientifically for matrix failures. Many of the contributions to the WWFE are a clear testament to that. The problem is that the most prominent of the bogus theories are promoted so intensely that the analysts and designers in this field are largely unaware of any need for improvements.
In conjunction with using only physically realistic models of composites in future, it is necessary to immediately cease teaching the innumerable interactive failure models based on the false assumption that it is permissible to replace the discrete fibre and resin constituents into an allegedly “equivalent” homogeneous anisotropic solid for any reasons other than to establish elastic constants or fibre-dominated laminate strengths. The many bogus theories should be taught only as examples of what NOT to do. In addition, it is necessary to expunge the bogus interactive theories from all the existing computer codes, to remove the temptation to use them because it takes more work to produce accurate answers. There is no need for them; there is nothing that the interactive composite failure theories based on homogenised anisotropic laminae can do “right” that the Truncated Maximum Strain model cannot do better.

What will it take to get the message through? Perhaps the makers of composite structures who have been bitten by unanticipated premature matrix failures, some of whom have suffered severe financial losses because of that, need to outlaw the further internal use of those theories that failed to warn them of the problems and to require that all of their subcontractors cease using those same theories on any composite products or components they provide. Otherwise, they will continue to suffer.

To those defenders of the status quo who still believe that the existing theories are adequate, the author would pose the following question. "Why not validate your position by comparing predictions with failure models that do not rely on artificial homogenisation to simplify the design and analysis process, or on arbitrarily changing the material reference properties to get a better answer?" Surely, this should have been done 40 years ago when these models were first proposed. And to the users of any theories who accept them blindly because other people already have, and they are easily available in computer codes; “Shouldn’t you first have understood what was in the models to know what they could cover and what they couldn’t?” This particular issue has far more widespread implications, particularly in the context of finite element analyses. The computer can, at best, analyse the structural (or whatever) model. It cannot correct for any omissions from the real situation – and it cannot provide any warning if the model was either incomplete or non-representative. Only experience can do that.

As the author has stated many times during the past 25 years, and will continue to say so until the appropriate people begin to listen, there is no such thing as a “composite material” – only composites of materials! The bottom line is that, once the fibres and resin have been mathematically homogenised, it is no longer possible to predict matrix failures.

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