Why fibrous composites are tough

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Abstract
From time to time there continue spectacular failures during the early service life of very high performance fibre composites. No comprehensive treatment of the toughness of fibre composites particularly under impact conditions is known to the authors. It seems worthwhile to recall the early experiments and ideas developed between 1965 and 1973 or so concerned with the physical principles of debonding, pull-out, transverse and longitudinal splitting, matrix plasticity, multiple cracking and interfacial shear all of which contribute to the toughness. Work associated with the names of Gordon, Cooper, Outwater, Tyson, Prewo, McGarry, Ewins, Potter, Bradshaw,Sidey and their colleagues as well as those of the authors and others will be recalled and commented upon. Finally some remarks will be made on the need for a diagnostic parameter for cumulative damage and fatigue of fibre composites and a suggestion made that this be measurement of Poissons ratio.

1.Introduction

Despite the current wide spread use of fibre composites (e.g. in the Boeing 787 where 80% of the volume and more than 50% of the weight of the airframe is made with fibre reinforced resin) there still occur from time to time spectacular failure of stressed composite components - many of which are not reported but some, which are, attract large attention (Philips catamaran, the tail fin and rudder of the American Airlines aircraft flight 587 in November 2001 and many formula one crashes).
It seems useful therefore to recall the early researches which studied how it is that composites can be tough and how in fact they can be engineered in order to give larger works of fracture per unit of weight than metal components.
At the time that very stiff fibres were invented there remained in the engineering memory the catastrophic failure of welded Liberty Ships during WW2 and the Comet disaster. It followed that the idea of a viable engineering material such as a composite of glass – the paradigm of a brittle material – combined with carbon, an equally brittle material, e.g. the lead in a lead pencil, appeared to run counter to common sense.
Engineering fibrous composites are usually used in the form of laminates. Within such a laminate the main in-plane tensile load is carried by the fibres aligned in the direction of the principal stress.

2.The crack tip

In a homogeneous material the relative importance of shear and tensile components may determine whether or not the material in question is inherently brittle or naturally ductile Kelly, Tyson and Cottrell [1]. A fibre composite is by definition inhomogeneous and the arrangement of fibre and matrix must be considered from the outset. The overall stress distribution is treated by conventional small strain elasticity taking into account the marked elastic anisotropy of the material

Accurate calculations, using anisotropic elasticity with elastic constants relevant to 50% carbon fibres in epoxy resin, show that, compared with the use of isotropic elasticity, there is a much larger concentrated shear stress parallel to the fibres and that the transverse tensile stress σ_x is much reduced relative to it. This indicates that shear failure is likely to be the most important mode. There is also a larger driving force for the crack to extend normal to the fibres.

If the crack is of depth c, the driving force necessary for extension normal to the fibres, producing so-called trans-fibre or trans-laminar failure, decreases inversely as the square root of c, whereas that for extension parallel to the fibres is independent of c. It follows, that for a deep crack, trans -laminar failure must be the observed mode even though the resistance to propagation along the fibres is usually much less than that for propagation normal to them.

3.Large and small notches

In the early years, the distinction between large (deep) notches and small ones was difficult to unravel. The distinction was clarified in the mid 1970s by work at Farnborough and other places; here we follow the discussion by Potter which one of us (AK) was pleased to have been asked to communicate to the Proceedings of the Royal Society.

Potter [2] argues as follows. Immediately ahead of a broken fibre, shear parallel to the fibres, accompanied, perhaps, by plastic flow and failure of the matrix and of the interface, will limit the stress concentration upon the first unbroken fibre . The concentrated applied stress varies across the section and so failure of the second fibre will only occur if, on failure of the first fibre under stress σ_f, the stress on the second fibre is within a small quantity $\Delta\sigma_f$ of this. Potter assumes that $\Delta\sigma_f$ is a characteristic constant for distinguishing large and small notches. If σ_2 is the stress on the second fibre when σ_1 (the stress on the first fibre) reaches σ_f we have as the condition of failure

$$\sigma_1 - \sigma_2 \leq \Delta\sigma_f = \frac{\partial\sigma_y}{\partial x}\frac{s}{N} \qquad (1)$$

Or
$$\left|\frac{\partial\sigma_y}{\partial x}\right| \leq \frac{N\Delta\sigma_f}{s} \qquad (2)$$

Here, s is the centre to centre separation of adjacent fibres and N relates the axial fibre stress to the laminate tensile stress resultant. σ_y is, of course, the local value of the stress, i.e. the stress at the notch tip, so that, if σ_{uc} is the composite tensile strength in the absence of the notch and σ_A is the applied stress, $\left|\frac{\partial\sigma_y}{\partial x}\right|$ must be evaluated when the applied stress is such that $\sigma_{uc} = K\sigma_A$ where K is given by a rather complicated expression for an orthotropic laminate which reduces to

$2(c/\rho)^{1/2}$ for isotropy, where ρ is the radius of curvature at the end of the notch. For a unidirectional laminate N is, of course, unity, but for other laminates is evaluated through normal laminate theory.
For 60% by volume of carbon fibres in epoxy resin Potter finds $\Delta\sigma_f \approx 2$MPa for a fibre spacing of 10microns. The average fibre strength is about 3.5 GPa so that $\Delta\sigma_f$ is about one-tenth of 1 percent of the fibre strength.

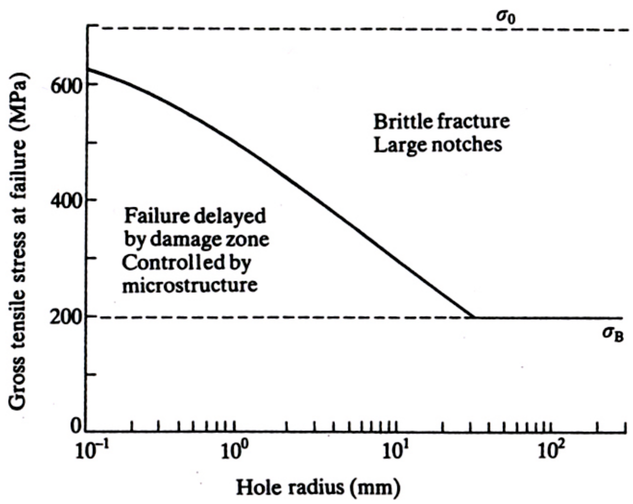

Figure 1 shows schematically the effect of holes on the strength of carbon fibre-reinforced epoxy laminate according to these ideas.
We have dealt with the behaviour of the crack as if the axial ply is the only one present. It might be supposed that the high in-plane shear stiffness due to the angled plies would modify any such criterion. However, load transfer between the fibres in an axial ply is dependent upon stress transfer via the matrix. Since the ply thickness is very much greater than the fibre spacing it follows that, even though the macroscopic in plane shear modulus may be considerably increased, any effect of stress transfer between two adjacent axial fibres will be negligible.

For a large notch the stress pattern, in particular the stress gradient at the point of maximum tensile stress in the y direction is such that laminate failure occurs as soon as the maximum tensile stress reaches the composite tensile strength. Small notches are characterised by the formation of stable damage zones which form at the end of the notch when the local tensile strength is reached. The resultant modification of the stress pattern due to the effects which we have described- largely obscures the effect of the initial shape of the notch. The crack is stable and will progress under an, initially, increasing stress as the fibres normal to the crack path are broken. The investigation of these damage zones within composite laminates is now a major activity of academic composite research. Known as Damage Mechanics it deals with the change under load of the size shape and number density of the cracks fissures and voids and debonded regions produced under stress or environmental change. The composite is then regarded as a continuum with changing microstructure and a phenomenological theory provides relationships between the overall stiffness and the severity of the damage -see for instance Talreja [3] and others
4.Varying resistance to crack propagation
The values of G_c for the tensile crack opening mode for the two cases when the crack runs parallel or normal to the fibres are very different, e.g. for a glass-

reinforced epoxy G_c for a crack running parallel to the fibre may be as small as 10^{-1} Jm^{-2} whereas for cracks normal to the fibres the value may be as large as 10^5 Jm^{-2}

The very different resistance to crack propagation in different directions and for different applied stresses means that the crack path in a fibrous composite under load may be unexpected. If a crack is proceeding parallel to aligned fibres under combined tension normal to the fibres and shear parallel to them, it may jump normal to the fibres. There are many other examples.

5.Contributions to the work of fracture

The values of G_{1c}, i.e. the resistance of the material to the propagation of a crack normal to the fibres, are extremely high for a fibrous composite; much higher than may be accounted for by adding the contributions of the two components. For example, the value for bulk glass is less than 10 Jm^{-2}. The value for a normal graphite is about 100 Jm^{-2}. In contrast the value for a glass reinforced with carbon fibres may be as large as 10^5 Jm^{-2} [4]. Clearly some interaction between the two components is responsible for this large increase. There are a number of contributions to the work of fracture: (a) the matrix may be plastically deformed as the crack is opened [5]); (b) broken fibres may be pulled out of their 'sockets' in the matrix on the other side of the crack [6]; and (c) an energy may be required to debond the fibre and matrix [7]. We deal with these in turn. Here and in what follows we use the usual notation, V = volume fraction, subscript m= matrix, f = fibre: r=fibre radius. τ is a shear stress representing a sliding friction between fibre and matrix.

The work done in deforming the matrix is proportional to the work done in deforming it to rupture per unit volume, U, times the volume of matrix deformed per unit area of crack surface. The volume deformed is equal to $V_m x'$, where, x' is equal to $(V_m/V_f)(r\sigma_{um}/2\tau)$ [5].

It follows that the total work of fracture due to this cause is proportional to $(V_m^2 r/V_f)$ and hence increases with increasing fibre radius. V_m must be greater than the critical value for effective reinforcement of course.

We can estimate W, the work done in pullout in separating the material into two pieces, when the stress distribution in each fibre is simply a linear build-up from the ends. In a discontinuous fibre compact, when fracture occurs by breaking of fibres, then it is found experimentally that all those fibres with ends within a distance $l_{crit}/2$ of the particular cross-section at which failure occurs, will pull out of the matrix instead of fracturing. The fraction of fibres pulling out will be (l_{crit}/l). The total work done, per unit area of specimen cross-section, in withdrawing all those fibres which pull out on both sides of the break is

$$W = \frac{V_f}{12}\left(\frac{l_{crit}}{l}\right)\sigma_{uf} l_{crit} \quad (3)$$

Cottrell [6] emphasised that to maximise the work of fracture l_{crit} should be made large and the fibre length should be kept closely equal to it. If $l>l_{crit}$ the work of fracture due to this cause decreases with increased length of the fibres. If $l< l_{crit}$ the work of fracture is $V_f\tau l^2/2r$ –using the definition of τ given above- and increases parabolically with fibre length. This variation of the work of pull-out with fibre length for a constant τ is experimentally found for discontinuous fibres [8]..

8.Debonding

A way of theoretically estimating the debonding energy from the properties of the components has been suggested by equating equate the energy of debonding with the elastic energy stored in the fibre after debonding. And this is the modern view [9]. If this is done, the work of debonding must always be much less than the total work of pull-out,

9.Multiple fracture of the matrix

When loaded in flexure or in tension most types of fibre composite show a load-elongation curve of the type shown in Fig 2. An initial elastic portion is followed after the point 1 by cracking of one component.

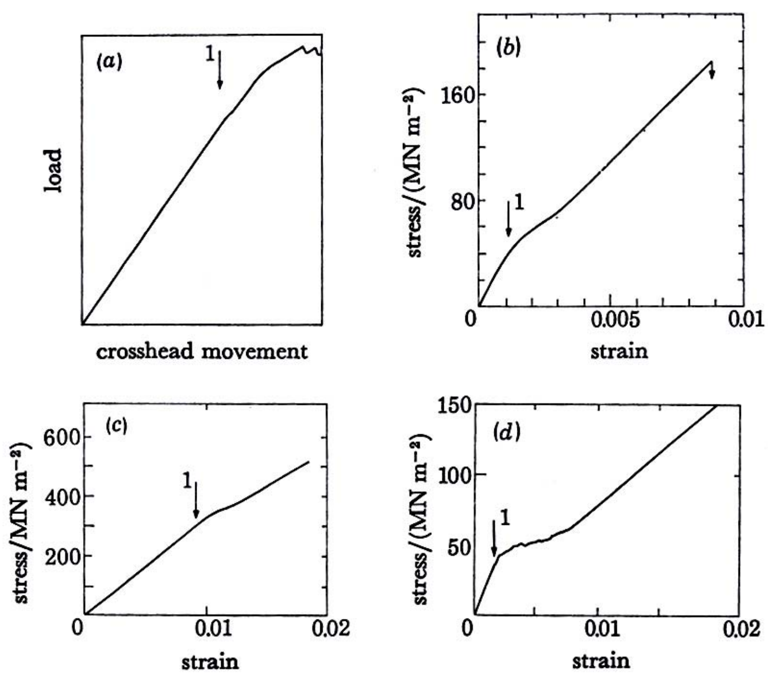

Fig 2

This may be the cracking of the 90^0 plies in a laminate (d), the cracking of the matrix if the fibres have the much greater strain to failure (b), or of the fibres if the reverse is the case (a), or the cracking of one set of fibres in a hybrid composite (c). The phenomenon was named multiple fracture and an elementary theory given by Aveston Cooper and Kelly[10] in 1971. A proper theory has been given by McCartney [11]) and by workers at Harvard [12].

Here we deal with the case when the matrix cracks.

If a specimen undergoes multiple fracture in all or part of its volume let the work done in the cracking process (including energy dissipated on both surfaces of the crack) be g_m per unit area. Hence, if there are $(1/x^/)$ cracks per unit of length, the total energy per unit volume will be $(g_m V_m/x^/)$. The usual value taken for $x^/$

assumes a linear build up of stress in the fibre characterised by a shear stress τ. We then have

$$x^{/} = \left(\frac{V_m}{V_f}\right)\frac{\sigma_{um} r}{2\tau} \qquad (4)$$

so in this case the energy dissipated increases with decrease in fibre radius. This part of the work of fracture is proportional to the total volume of the specimen which undergoes multiple fracture. After the process of cracking of the matrix is complete, further work is done in deforming the specimen to failure. This is easily found from the stress-strain curve when this curve is of the form of that in Fig.2. The total work done, W, is given by

$$W = (1/2)\sigma_{uf}\varepsilon_{uf}V_f + \eta\alpha E_c \varepsilon_{um}^2 \qquad (5)$$

with $\alpha=(E_m V_m/E_f V_f)$

where η is a constant between 0.125 and 0.159, depending on the precise assumption made concerning the average value of the crack spacing. Multiple fracture, together with pull-out is the major contribution to the work of fracture of fibre-reinforced ceramics e.g.[13] and cement [14]

The three principal methods of dissipating energy and delaying crack advance are then; pull-out, debonding and multiple fracture. It may be shown [15] that the work of debonding is always much less that that due to pullout of the fibres

We may compare the values obtained by pull-out with those found in other tough high strength materials. Using (3) for CFRP of V_f =0.5 with l_{crit} =50μm, and (l/l_{crit})=0.75, σ_{uf}=3GPa we have $W \approx 4.5 \times 10^3 J/m^2$ for a material with a breaking strength of ≈1.5GPa.This value compares with say, a high strength maraging steel of similar strength, but, of course of very much higher density, which would have a value of about $7 \times 10^3 J/m^2$- hence of much the same order.

Values for the energy dissipated by multiple fracture are not expressible as a conventional strain energy release rate in fracture mechanics because the whole specimen is involved and the work done is not concentrated near the fracture surface. A striking example of the release of stress concentration at the root of a notch in a very brittle material is shown in Fig 3 due to Carl Cady.

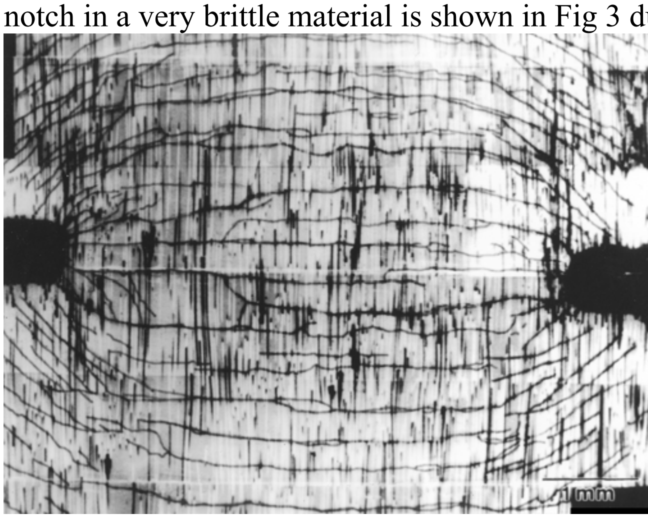

Fig 3.Cracking in a notched specimen of 40% SiC fibres in a calcium aluminosilicate glass.

10 Comparing metals and composites

A proper comparison of the toughness of a composite material compared with a metallic structure can only truly be made if specimens of the same shape in the two types of materials are deformed to fracture. Table 1 compares Energy Absorbed in Crushing Tubes of Various Materials of the type used to provide crash worthiness, e.g. in helicopters. In glass and in carbon reinforced epoxies and in polyester matrices energy is absorbed by multiple microfracture processes. These processes are much more efficient in absorbing energy than those occurring in metallic structures.

Table 1Energy Absorbed in Crushing Tubes of Various Materials [16]

Material	Speed of Crosshead ($mm\ s^{-1}$)	Specific Energy (Jg^{-1})	Energy/volume (Jml^{-1})	Density (Mgm^{-3})
Polycarbonate	4000	15	24	1.2
PVC	4000	brittle	-	1.4
Mild steel	4000	25	195	7.8
Aluminium	-	16	43	2.7
Composite 0/90 polyester	4000	55	104	1.9
SMC	4000	37	67	1.8?
Glass vinyl ester	-	70	133	1.9
Carbon -epoxy	2000	110	176	1.6

This account has outlined in a semi quantitative manner how it is that fibre composites can be extremely tough and show adequate values of energy absorption comparable to those obtained with metals. We have not dealt with fracture under impact conditions but here again composite materials

Table 2

	Energy kJm^{-2}
Carbon Fibre untreated in Epikote 828	300
E glass in Epikote 828	580
Aluminium RR58	540

show comparability. Table 2 shows values of the Izod impact energies for unnotched specimens of CFRP, GFRP and an aluminium alloy taken from early work at Farnborough by Bradshaw, Dorey and Sidey

The spectacular failures which occur in practice are sometimes due to misuse as in the case of American Airlines flight 587 where the manoeuvring of the aircraft led to the design loads being exceeded – but may continue unless toughness is built into the computer codes needed to define the properties. These deal well with elastic properties but are crude in their treatment of toughness and failure if they deal with it at all.

11.Fatigue of Composite Laminates

We have emphasized that for most laminates in service the effect, under load, of notches or of sharp changes of section is understood. Changes during service are often described in terms of the so-called damage mechanics- and the specimens usually undergo fatigue.

A composite consisting of aligned high stiffness fibre undergoing cyclic loading parallel to the fibres shows little fibre breakage [17] and very small decrease in stiffness in the axial direction because the damage occurs at the fibre matrix interface- hence parallel to the load. In the case where the failure strain of the fibres is less than that of the matrix little fatigue damage occurs at loads less than 80% of the breaking load and it is this type of behaviour which has led to the statement that high performance composites show great resistance to fatigue damage. However this is only true in this special case. In general the nature of the damage produced by cyclic loading is complicated because many types of damage can be observed, for instance: film fracture, matrix cracking, fibre buckling, longitudinal splitting, ply delamination etc. Even for a precisely defined composite structure it is usually not known which type of damage is the most significant..

The picture which is generally accepted and used for metals, is that the total fatigue life (or number of cycles to failure) is composed of (i) the number of cycles required to initiate a fatigue crack in a nominally defect (or crack) free specimen plus (ii) the number of cycles required to propagate the dominant crack to final failure. The growth of these cracks obeys a rather well defined growth law, originally due to Paris in 1961 namely

$$(da/dN) = C(\Delta K)^m \qquad (6)$$

where N is the number of cycles, ΔK the range of stress intensity factor and C and m are material constants, influenced by microstructure, yield strength and environment. A consequence of the form of this equation and experimental values of m (always greater than 1) is that the largest crack in a population grows much more rapidly than others and determines the life. Total life estimates derived from observation on laboratory specimens with smooth surfaces, then primarily denote the life to initiate a fatigue crack, because the crack initiation component of total life in such cases is often 80%. So a tenable fatigue life philosophy can be interpreted as focussing on design against fatigue crack initiation in smooth surfaced specimens.

Because of the many uncertainties with composites composite designs for specific components are grossly “overdesigned” so that stresses are greatly reduced. This results in an artefact in a composite material becoming grossly overweight and over priced compared with the putative competitor materials. We have personally seen this happen where fibre composite materials made with higher temperature capability resins, could replace titanium alloys in aeroengine components were it not for the fact that the uncertainty of behaviour of the former material under fatigue conditions, compared with the latter leads to excessive “safety” or “uncertainty” factors being applied to the composite material.

We suggest an observable parameter for fibre composite materials .
In contrast to the metal case where fatigue is not accompanied by a marked change in the elastic constants, composite laminates deteriorate through breaking of fibres, shear, debonding and the formation of cracks within the material. There is therefore a decrease in density and an increase in permeability and a loss of stiffness. The phenomenological study of these phenomena is, as we have said, the subject of Damage Mechanics. The loss of stiffness can be monitored and we suggest that the measurement of Poissons ratio could be used as an index of the damage. In a 0/90 composite the decrease in Poissons ratio with increase in damage is much more marked than the loss of tensile or of shear stiffness[18].

An elastically orthotropic laminate possesses 3 independent values of Poissons ratio-an example is shown in Figure 4 for high performance carbon fibre in epoxy resin.

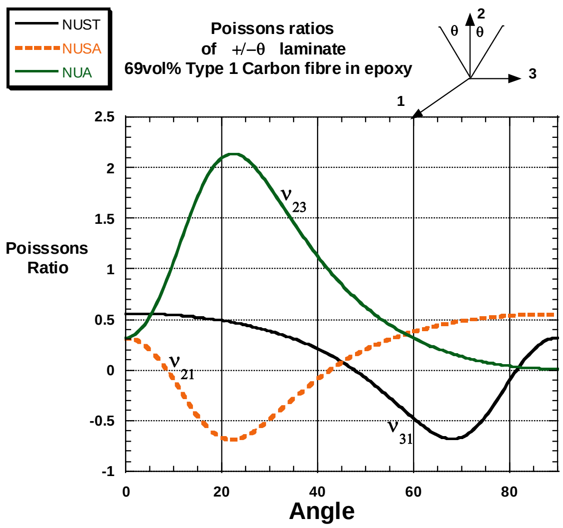

Fig 4
It is an experimental fact that Poissons ratio is much more sensitive to damage produced during fatigue testing than are the other elastic constants[18]. And the change in value is also well predicted by methods of Damage mechanics. Monitoring of the value of Poisson's ratio is relatively easy and non-destructive. It may be argued the the principal Poissons ratio is often rather small, and hence difficult to measure, but this is only true of a particular orientation –angle 0 in Figure 4. Angle ply laminates are widely used in many structures and in filament wound structures are paramount. They show extremely large values of the axial Poissons ratio- Figure 4- often exceeding 1.5 and hence are ideal for monitoring.
A typical glass reinforced plastic laminate with a stacking sequence often used in practice and called quasi isotropic with symmetric sequence 90 0 -45 +45would gives values shown in the Table below. All the values would be easily measurable

Direction	ν_{23}=nuA, axial value	ν_{21}=nusa	ν_{31} =nust through thickness
Value	0.314	0.365	0.365

Direction 1 is parallel to the axis (0^0 direction of the fibres) 2 in-plane normal to this and 1 is through the thickness (ttt)

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