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Failure of Fibrous Composites. Part 1
Stress Concentrations, Cracks and Notches
Part 2: Toughness

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FAILURE OF FIBROUS COMPOSITES

Part 1 Stress Concentrations, Cracks and Notches

Part 2 Toughness

Peter W.R. Beaumont

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Stress Concentrations, Cracks and Notches (M190/00298)

In designing structural components from brittle fibrous composites, glass fibres in epoxy, for instance, we assume that the operating stress does not exceed the strength of the composite for an acceptable level of survival probability. Unfortunately, it is not so straightforward. A fibrous composite containing a notch or hole in monotonic loading exhibits premature cracking at stresses significantly lower than the ultimate strength; interfacial shear cracking, delamination or splitting, fibre fracture, matrix cracking and so forth. The formation of a damaged region is due to the concentration of localised tensile and shear stresses close to the notch front. The precise mode of failure depends upon the orientation of fibres and stacking geometry of the laminate; it is also sensitive to the properties of the matrix and fibre-matrix interface; and to the stress-state and environment.

Failure Modes in Monotonic Loading

A typical unidirectional fibrous composite, glass fibres in epoxy, loaded in tension exhibits stable delamination or splitting at the root of a notch. As the split extends parallel to the fibres and direction of applied load, a matrix crack propagates from the notch tip, passing fibres still intact. A second split nucleates at the crack tip and the crack becomes arrested. Further crack growth in the original notch orientation requires extension of the two splits followed by fracture of the intact fibres somewhere along their debonded length. The extent of delamination, fibre debonding and fibre fracture in the localised damage zone is related to the applied load. We can think of the split which forms as rendering the sharpest inherent defect or crack effectively equal to a hole or notch (Mandell, 1971).

In a monotonic tensile test on a cross-ply ($0^o/90^o$) laminate, the transverse (90^o) layers hinder longitudinal splitting at the root of a notch. Those layers perpendicular to the direction of applied load and adjacent to an interfacial shear crack may also delaminate slightly during shearing between the two sliding surfaces of a longitudinal split. Premature fracture of the localised (0^o) load-bearing fibres may lead to subcritical crack growth in the original notch direction. Behind the crack front lies an array of longitudinal, parallel shear cracks which mark the position of successive crack arrest points. The length of the delaminations and spacing between them are related in some way to the laminate stacking geometry.

For a quasi-isotropic, laminate ($0^\circ/\pm45^\circ/90^\circ$), delamination in the ($45^\circ$) layers at the root of a notch reduces the concentration of stress on the (0°) plies. An increase in thickness of the (45°) layer decreases the constraining effect on the (0°) plies which permits further delamination (Bishop and MacLaughlin, 1979). We can think of the formation of a damage zone at the notch tip having a lower modulus than the surrounding material, and the intensification of localised stress decreases. In effect, the notched strength and fracture toughness of the laminate is increased. For some, perhaps all laminates containing glass, Kevlar or carbon fibres, we can think of the damage zone increasing the effective radius ρ of the notch (Potter, 1978). The fracture stress of a notched composite can then be estimated using measurements of ultimate strength, σ_u, together with a suitable stress concentration factor:

$$K_t = \frac{\sigma_u}{\sigma_f} = 1 + \left(\frac{c}{\rho}\right)^{\frac{1}{2}} \left[2 \left(\frac{E_{11}}{E_{22}}\right) \nu_{12} + \left(\frac{E_{11}}{G_{12}}\right)\right]^{\frac{1}{2}}$$

At the microscopic level, the breakage of a fibre at the tip of a notch may induce the sequential failure of longitudinal (0°) fibres by the transfer of load from the broken fibre to an adjacent intact fibre. The localized concentration of stress on the unbroken fibre is sensitive to the strength of the fibre-matrix interface and matrix. For example, we would expect a toughened matrix, epoxy dispersed with elastomeric spheres, to reduce the amount of delamination at the notch front, and this is observed. However, reducing splitting raises the localised stress at the notch tip and the fracture toughness corresponding to crack propagation perpendicular to the (0°) fibres is lowered.

The initiation of cracking parallel to the original notch direction may be prevented if, as a result of the stress distribution due to the notch, the initial difference in stress carried by a fibre which has just broken and the adjacent intact fibre is sufficiently large. The interaction between the distribution of stress in the vicinity of a notch tip and the concentration of localised stress in an unbroken fibre next to the one that has just failed leads to a notch size effect on strength and fracture toughness (Potter, 1978). The smaller the notch, the more localised becomes the perturbed stress field, and the greater becomes the applied stress to initiate fibre fracture and transfibrillar crack propagation. It is this subtle balance between the transfer of load to an intact fibre next to a broken fibre, the distribution of flaws in the fibre and the variability of fibre strength, combined with the localised stress field ahead of a notch tip that determines the strength and notch sensitivity of the laminate.

We can relate fracture touchness K_c to the damage zone size C_o and notch geometry (Nuismer and Whitney, 1975):

$$K_c = \sigma_u \left[\pi \ (C + C_o)(1 - \xi^2)\right]^{\frac{1}{2}}$$

where $\xi = C/(C + d_o)$; d_o is a critical distance ahead of the discontinuity.

The concept of a damage zone is invoked so that K_c reaches a maximum value as the notch length C becomes very small - $K_c \rightarrow \sigma_u (\pi C_o)^{\frac{1}{2}}$. Typical values of K_c, σ_u and C_o for various laminates and stacking geometries are listed in Table 1.

Cyclic Failure and Residual Strength

Load cycling of fibrous composites containing brittle fibres, carbon fibres or glass fibres in epoxy, for instance, brings about fibre-matrix decohesion and delamination at notches and inherent flaws.

These damage zones are larger than induced in a monotonic loading experiment. The nucleation and growth of damage zones in cyclic failure reduces the localised stress, together with a corresponding improvement in residual strength. The residual strength of a laminate depends on the stress level and increases with time in a cyclic loading experiment, eventually reaching the unnotched or inherent strength in monotonic fracture. Fatigue lifetime and residual strength, (the two are inseparable), are therefore affected by interactions between microstructure, distribution of flaws, and the formation of damage zones, shear cracking and so forth. The micromechanisms of failure are sensitive to the chemistry of the resin and nature of the fibre-matrix bond. Extrinsic variables, temperature, humidity, and time in an aging experiment, for instance, are likely to have considerable influence on the properties of the fibre-matrix interface and therefore upon residual strength and fatigue lifetime of the laminate.

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TABLE 1

TYPICAL STRENGTH AND FRACTURE TOUGHNESS DATA

Laminate Construction	Tensile Strength (σ_u) (MN/m^2)	Measured Fracture Toughness (K_c) ($MN/m^{3/2}$)	Estimated Damage Zone Size (C_o) (mm)
HTS Carbon Fibre/Epoxy			
(0)	1,045	90.0	1.86
$(0/90)_{2S}$	530	42.1	1.78
$(0/\pm45)_{2S}$	460	33.2	1.02
$(0/\pm45/90)_{2S}$	270	32.8	1.49
E-Glass Fibre/Epoxy			
(0)	1,035	-	-
$(0_2/\pm45)_{2S}$	600	-	-
$(0/90)_{2S}$	540	30.7	-
$(0/\pm45/90)_{2S}$	350	24.3	2.54
Boron Fibre/Epoxy			
(0)	1,325	-	-
$(0_2/\pm45)_{2S}$	700	67	2.54
$(0_2/\pm45/90)_{2S}$	420	38.7	2.79
Kevlar 49 Fibre/Epoxy			
(0)	1,378	-	-
$(0/90)_{2S}$	578	-	-
$(0/\pm45/90)_{2S}$	393	∿ 30	-
(±45)	119	∿ 14	-
Boron Fibre/Aluminium			
(0)	∿1,000	90	-
$(0/\pm45)_{2S}$		50	-

Toughness (M190/00299)

This article surveys the micromechanisms of crack extension in various classes of fibrous composites and includes models to account for the origins of toughness. The subject can be approached in two complementary ways: empirically, by assembling a large amount of toughness data for a given composite system; and theoretically, using models for the individual fracture mechanisms. The models can be used to distinguish between the micromechanisms of fracture; give guidance in applying toughening methods; and assist in selecting a fibre/matrix system for a particular application.

Broad Classes of Fracture Mechanism

When a strong fibre is dispersed in a ductile or brittle solid, a multiplicity of quite different micromechanisms of fracture can occur. The broad classes of fracture mechanism are illustrated in Fig 1. In a brittle fibrous composite system, glass fibres in epoxy, for instance, at least four distinguishable mechanisms are possible, each event contributing to the toughness. They are decohesion of the fibre-matrix interface (fibre debonding), fibre fracture, matrix cracking and fibre pull-out. A brittle fibre-ductile matrix system, boron fibres in aluminium, for example, can fail by fibre fracture followed by *localised* plastic flow in the matrix adjacent to the broken fibre ends.

It would be useful to have some idea of the conditions under which each process appears, and how the toughness of the composite might change, if, for example, the matrix was toughened, the fibre surface treated to improve the strength of the fibre-matrix bond, or the ductility of the matrix increased.

Micromechanisms of Fracture; The Origins of Toughness

First, we present models of fracture, together with equations to quantitatively describe the origin of toughness. Most are based on published work, and all are simplified while still retaining the essential physics of the fracture mechanism. Frequently, this leads to an expression containing one or more terms for which only bounds are known or can be found. Theory provides only the form of the equation. These models together with experimental and theoretical values of toughness of various fibre/matrix systems, are summarised in Table 1.

Fibre debonding

A convenient point which provides the basis for most of the micromechanisms of fracture and toughness is the determination of shear stress at a fibre-matrix interface when a fibre is loaded in tension. The stress required to initiate decohesion and to extend a debonded region of fibre is, (Outwater and Murphy, 1969),

$$\sigma_d = (8 E_f \mathcal{G}_c / d)^{1/2} + 4 \tau \ell_d / d$$

τ is the sliding shear stress set up soon after the bond has failed; ℓ_d is the distance over which decohesion has occurred; E_f is fibre modulus; and $\mathcal{G}_c$ is the toughness.

Provided $d < E_f \mathcal{G}_c / \sigma_f^2$, (σ_f is fibre strength) then the fibre will snap. The toughness is given by

$$\boxed{\mathcal{G}_c = \sigma_f^2 d / 8E_f}$$

Values of $\mathcal{G}_c$ can range between 6 J/m^2 for glass fibres in cement and $10^3 J/m^2$, approximately, for glass fibres in epoxy.

Post-Debond Fibre Sliding

An additional contribution to the fibre debonding energy originates from the sliding action of the fibre in its matrix sheath. A frictional force of $\tau \pi d (\ell_d/2)$ is established soon after the bond has failed, acting in both directions from the fracture surface of the matrix. The distance over which sliding occurs is equal to the difference in displacement between fibre and matrix and is of the order of $\ell_d (\epsilon_f - \epsilon_m)$. The work done per fibre is given by, (Kelly, 1970),

$$W_{pdf} = [\pi d \tau \ell_d^2 (\epsilon_f - \epsilon_m)]/2 ,$$

and

$$\boxed{\mathcal{G}_c = 2 \tau \ell_d^2 (\epsilon_f - \epsilon_m)/2}$$

Values of $\mathcal{G}_c$ can range between $10^2 J/m^2$ for carbon fibres in epoxy and 10^5 J/m^2 for glass fibres in epoxy.

Fibre Breakage and Stress Relaxation

Under increasing load conditions, the fibre eventually snaps, dissipating deformational (elastic) energy. The load is redistributed over a critical length, $\ell_c/2$, from the broken fibre ends. For simplicity, the stress in the fibre can be considered to increase linearly from the end of the fibre. The energy released by a fibre snapping is, (FitzRandolph, Phillips, Beaumont and Tetelman, 1972),

$$W_r = \pi d^3 \sigma_f^3 / 24 E_f \tau ,$$

and

$$\boxed{\mathcal{G}_c = \sigma_f^3 d / 6 E_f \tau}$$

Values of $\mathcal{G}_c$ can range between about 10^3 J/m² for glass fibres and carbon fibres in epoxy and 10^4 J/m² for boron fibres in epoxy.

Fibre Pull-Out

Following fibre fracture and matrix cracking, the fibre is extracted from the matrix sheath. Any frictional shear force at the interface opposes a force applied to the fibre. The average work to pull out a fibre whose embedded length lies between 0 and ℓ_p beneath the fracture plane of the matrix is, (Kelly, 1970),

$$\bar{W}_p = \left(\frac{1}{\ell_p}\right) \int_{\ell=0}^{\ell=\ell_p} W_p \, d\ell ,$$

and

$$\boxed{\mathcal{G}_c = 2 \tau \ell_p^2 / 3 d}$$

Values of $\mathcal{G}_c$ are of the order of 10^3 J/m² for glass fibres in epoxy; 10^4 J/m² for carbon fibres in epoxy; and 10^{3-4} J/m² for carbon fibres in aluminium.

In the case of a discontinuous (short fibre) composite, fibres of length $\ell < \ell_c$ will always pull-out during composite fracture. For fibres of length $\ell > \ell_c$, those fibres whose ends lie within a distance $\ell_c/2$ of the matrix crack surface will pull-out; the remainder will be stressed to their breaking point and fracture. The average work to pull out a short fibre is, (Kelly, 1970),

$$W_p = \int_0^{\ell/2} \pi d \tau \ell^2 \, d\ell / 2 \int_0^{\ell/2} d\ell ,$$

and

$$\boxed{\mathcal{G}_c = \tau \ell^2 / 6 d} \qquad \boxed{\ell < \ell_c}$$

The limit $\ell/2$ is chosen rather than ℓ since a fibre spanning 10
a matrix crack will always pull out from a fracture surface beneath which the fibre is least buried. Where fibres are longer than ℓ_c, only a fraction ℓ_c/ℓ, of them will be extracted, the remainder failing in the plane of the matrix crack. The average work of pulling out a fibre is

$$\bar{W}_p = \left(\frac{\ell_c}{\ell}\right) \int_0^{\ell/2} \pi d \tau \ell^2 / 2 \int_0^{\ell_c/2} d\ell \, ,$$

and

$$\boxed{\mathcal{G}_c = \tau \ell_c^{\,3} / 6 \ell d} \qquad \boxed{\ell > \ell_c}$$

Values of $\mathcal{G}_c$ are of the order of $10^3 - 10^4$ J/m² for carbon fibres in Nylon.

For a composite whose fibre length is equal to ℓ_c,

$$\boxed{\mathcal{G}_c = \sigma_f^{2} d / 24 \tau} \qquad \boxed{\ell = \ell_c}$$

In physical terms, the work of fibre pull-out is dependent upon fibre diameter and inversely proportional to the frictional interfacial shear stress. A maximum in toughness occurs when $\ell = \ell_c$.

For a composite containing long fibres, the inference is that the toughness falls to zero as $\ell \to \infty$. This is not true in practice; an epoxy containing long carbon fibres, for example, has a toughness of the order of 10^4 J/m². It is the variation in strength of the fibre along its length and the fracture of one of the weak points beneath the surface of the cracked matrix, followed by fibre pull-out which is responsible for the high toughness.

Plastic Deformation of Matrix (Non-Ductile Fibres)

In this micromechanism, account is taken of the contribution of the matrix to the toughness of the composite. The fracture plane consists of a plane of fibre fractures and between them a series of matrix "bridges" which dissipate eneryg during plastic flow . At failure, the maximum stress carried by the matrix is $\sigma_m(1-V_f)$, (V_f is the fibre volume fraction). The distance over which this stress is transferred to the fibre is, (Cooper and Kelly, 1967),

$[(1-V_f)/V_f][\sigma_m d / 4 \tau_m]$. The toughness is therefore equal to

$$2x(1-V_f)\int_0^{\epsilon_m} \sigma_m \, d\epsilon \, ,$$

or

$$\mathcal{G}_c = [(1-V_f)^2/V_f]\, d\, \sigma_m \epsilon_m$$

Toughness depends upon fibre diameter, matrix yield stress, σ_m, and its ductility, ϵ_m. A sharp increase in toughness occurs at low values of V_f. Values of $\mathcal{G}_c$ are about 10^2 J/m² for boron fibres in aluminium.

Matrix Flow Around Multiple-Fibre Breaks (Brittle Fibres)

This model assumes the breakage of many brittle fibres ahead of a propagating crack. An example might be boron fibres in aluminium. A ductile matrix flows easily in a multiple-fibre fracture zone. The size of this damaged region depends upon ℓ_c. The volume of damaged material per unit area of crack surface is $2(1 - V_f)\, \beta\, \ell_c$, where β is some multiple of the critical fibre length ℓ_c. In terms of the work of plastic shearing, the toughness is, (Gerberich, 1971),

$$\mathcal{G}_c = \beta\, \epsilon_m\, (1-V_f)\, \sigma_f\, d$$

Values of $\mathcal{G}_c$ are of the order of 10^5 J/m² for boron fibres in aluminium.

Plastic Deformation of Fibres (Brittle Matrix)

In this case, the toughness is equated to the work done in plastically shearing the fibre as it is extracted from a cracked matrix, a random array of short steel wires in concrete is an example. Those fibres which are misoriented with respect to the applied tensile stress direction will straighten out by a plastic shearing process, provided $\ell/d < \epsilon_m E_m/\tau V_f$. The work to plastically deform the fibre is $f(V_f, \ell, \theta) \int_0^{\epsilon_f} \sigma_f\, d\epsilon$. θ is the angle through which the fibre undergoes plastic flow over a distance of the order of d/2 on either side of a matrix crack. The expression for toughness has the form, (Tetelman, 1969),

$$\mathcal{G}_c = \sigma_f\, \epsilon_f\, 2\, d\, V_f$$

For the first case, where $\ell/d < \epsilon_m E_m/\tau V_f$, an oblique fibre to the matrix crack plane has to do additional work to the work of fibre pull-out, the work to pull the fibre around the corner of the fibre socket and matrix crack plane. Values of $\mathcal{G}_c$ are of the order of 10^3 J/m² for steel wires in concrete.

Interfacial Shear Cracking and Splitting

When a portion of a composite between two neighbouring splits or interfacial shear cracks in the vicinity of a notch fails, the entire portion of longitudinal ply partially unloads and retracts (see article on Cracks and Notches). The process of retraction dissipates stored elastic strain energy as a crack propagates from one split to the next and

$$\mathcal{G}_c \approx \frac{1}{2}(\overline{\Delta\sigma})(\overline{\Delta\epsilon})\ell$$

$\overline{\Delta\sigma}$ and $\overline{\Delta\varepsilon}$ are the average decrease in stress and strain along the split, and ℓ is the length of split. As a first approximation, $\overline{\Delta\sigma}$ and $\overline{\Delta\varepsilon}$ are equal to one-half of the tensile strength and failure strain of the longitudinal ply. For a 25 mm split in a $(0^\circ/90^\circ)$ glass fibre-epoxy laminate, $G_c \approx 10^5$ J/m^2 which is of the same order of magnitude as the measured toughness.

Final Comments

Fibrous composites can be classified by their fracture behaviour. For each composite system, a model can describe the principal mechanism of fracture: fibre fracture, matrix shearing, fibre pull-out and so forth. Order of magnitude calculations of G_c based on the micromechanisms allow a mechanism to be associated with the toughness of a particular composite.

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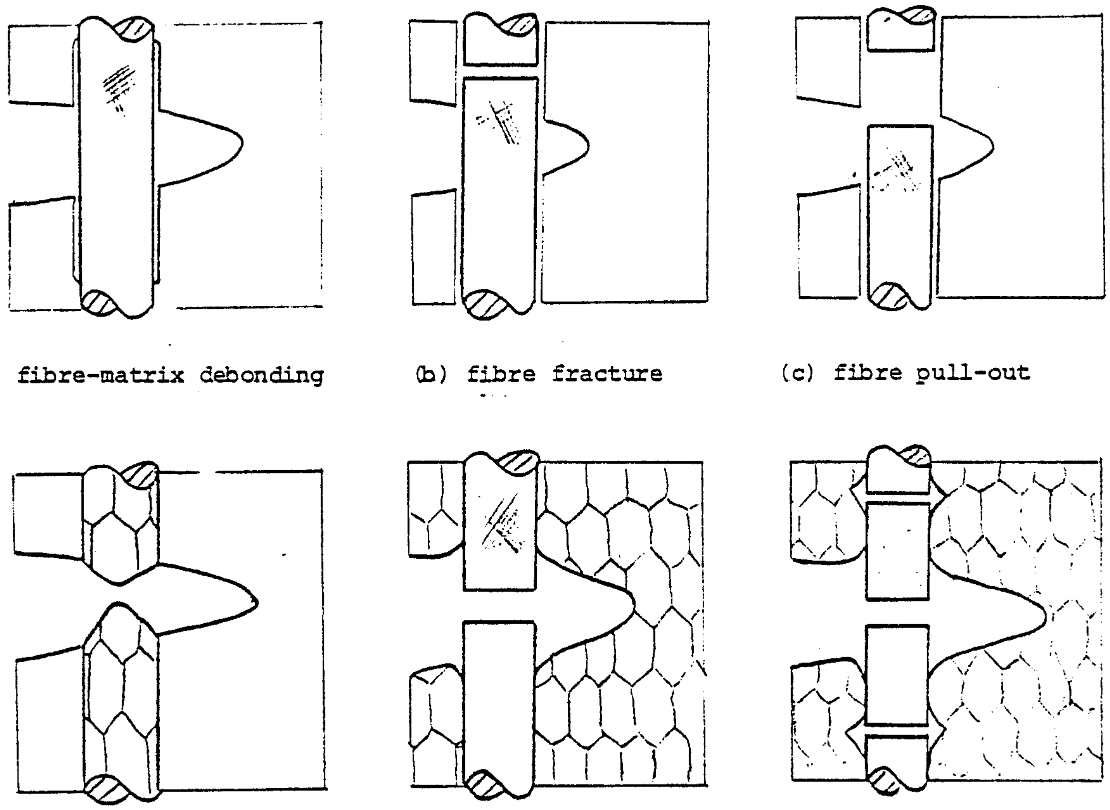
TABLE 1 Magnitudes of G_c and their relationship to the mechanism of crack extension for typical unidirectional fibrous composites.

MECHANISM	EQUATION	TOUGHNESS, G_c (J/m^2) GLASS/EPOXY	CARBON/EPOXY	BORON/ALUMINIUM
Fibre-matrix debonding	$G_c = \sigma_f^2 \, d/8 \, E_f$	10^3	10^2	-
Fibre fracture	$G_c = \sigma_f^3 \, d/6 \, E_f \, \tau$	10^3	10^2	10^3
Fibre-matrix sliding	$G_c = 2\tau \, \ell_d^2 \, (\varepsilon_f - \varepsilon_m)/2$	10^5	10^3	-
Fibre pull-out	$G_c = 2\tau \, \ell_p^2/3d$	10^3	10^4	-
Matrix shearing	$G_c = \lvert(1 - V_f)^2/V_f\rvert \, d \, \sigma_m \, \varepsilon_m$	10^2	10^2	10^3
Multiple fibre fracture and matrix shearing	$G_c = \beta \, \varepsilon_m \, (1 - V_f) \, \sigma_f \, d$	-	-	10^5
Interfacial shear cracking and splitting	$G_c = \frac{1}{2} \, (\overline{\Delta\sigma})(\overline{\Delta\varepsilon}) \, \ell$	10^5	-	-
TYPICAL EXPERIMENTAL VALUES*		$\sim 10^5$	$2\text{-}5 \times 10^4$	$\sim 10^5$

*(For comparison, Kevlar 49/Epoxy has a toughness, G_c, of the order of 10^5 J/m^2).

FIG. 1 BROAD CLASSES OF FRACTURE MECHANISM



(a) fibre-matrix debonding

(b) fibre fracture

(c) fibre pull-out

(d) rupture of fibre

(e) matrix flow around fibre break

(f) matrix flow around multiple fibre breaks

The simplest classification of fracture mechanism. The upper row shows a sequence of failure events under increasing load; (a) fibre-matrix debonding, (b) fibre fracture and slippage, and (c) fibre pull-out. The lower row shows failure events in which (d) the fibre necks down and ruptures, (e) the fibre fractures and the matrix necks down and ruptures, and (f) the fibre fractures at several points and the matrix flows and ruptures close to each fibre break.