8
Properties of Composite Systems

8.1 Introduction

The mechanical properties of simple unidirectional continuous fiber composites depend on the volume fraction and properties of the fibers (including flaw and strength distribution), the fiber/matrix bond strength, and the mechanical properties of the matrix. The alignment (waviness) of the fibers also has a significant effect on some properties—notably, compression strength. Elevated temperature and moist environments also significantly affect properties dependent on matrix properties or interfacial strength.

Because of these and several other factors, the efficiency of translation of fiber properties into those of the composite is not always as high as may be expected. Stiffness can be predicted more reliably than strength, although static tensile strength is easier to predict than other strength properties. Chapter 2 provides some elementary equations for estimating the mechanical properties of unidirectional composites, which are reasonably accurate in estimating elastic properties, providing fiber alignment is good. The equations can also provide ball-park figures for the matrix-dominated shear and transverse elastic properties and the fiber-dominated tensile strength properties. However, estimation of matrix-dominated or fiber/matrix bond strength–dominated strength properties, including shear and compression, is complex. Prediction problems also arise when the fibers are sensitive to compression loading, as is the case for aramid fibers, as discussed later.

Chapter 7 describes the experimental procedures for measuring the mechanical properties, including those for assessing tolerance to damage and fatigue. These tests are used to develop a database for design of aerospace components and as part of the information required for airworthiness certification, as described in Chapter 12.

Table 8.1 lists relevant mechanical and physical properties of the composites discussed in this chapter. Details of aerospace structural alloys aluminum 2024 T3 and titanium 6Al4V are also provided for comparison. The nomenclature used for the properties is similar to that used in Chapter 2. The data provided for the composites can be used as an estimate of ply properties for making an approximate prediction of laminate properties.

The first four sections of this chapter provide an overview of the mechanical properties of composite systems based on glass, boron, aramid, or carbon fibers.

Table 8.1 Unidirectional Properties (Mostly Approximate) of Various Composites Considered in This Chapter and, for Comparison, Airframe Titanium and Aluminum Alloys. Based Largely on Ref. 2

		Glass fiber composites				Carbon fiber composites				
	Units	E	S	Boron	Aramid K 49	HT	HM	UHM	Al	Ti
SG	—	2.1	2.0	2.0	1.38	1.58	1.64	1.7	2.76	4.4
α_1	μm/°C	7.1	6.3	4.5	−1	−0.16	—	—	23	9
α_2	μm/°C	20	—	—	70	24	—	—	23	9
σ_{1tu}	MPa	1020	1620	1520	1240	1240	760	620	454	1102
σ_{1cu}	MPa	620	690	2930	275	1100	690	620	280*	1030*
σ_{2tu}	MPa	40	40	70	30	41	28	21	441	1102
τ_u	MPa	70	80	90	60	80	70	60	275	640
ILS	MPa	70	80	90	60	80	70	60	—	—
E_1	GPa	45	55	210	76	145	220	290	72	110
E_2	GPa	12	16	19	5.5	10	6.9	6.2	72	110
G_{12}	GPa	5.5	7.6	4.8	2.1	4.8	4.8	4.8	27	41
ν_{12}	—	0.28	0.28	0.25	0.34	0.25	0.25	0.25	0.33	0.31
ε_{1u}	—	0.022	0.029	0.006	0.016	0.01	0.03	0.02	0.12	0.06
ε_{2u}	—	0.4	0.4	0.4	0.5	0.4	0.4	0.3	—	—

Notes: Ti ≡ Ti 6Al4V; Al ≡ 2024 T3 *yield value

V_f, ~60% in the composites; SG, specific gravity; ILS, interlaminar shear. See Chapters 2 and 12 for definition of other terms.

Further information on the properties of carbon-fiber composites is also provided throughout this book. In the last three sections, more generic discussion is provided on important impact, fatigue, and environmental properties, while a focus on carbon-fiber systems is maintained.

8.2 Glass-Fiber Composite Systems

As described in Chapter 3, several types of glass reinforcements are suitable for the manufacture of aircraft and helicopter composite components. E-glass composites are used extensively in gliders and in non-structural components that do not require high stiffness, such as radomes. S-glass composites have better mechanical properties and therefore are used in more demanding applications. A third type of reinforcement known as D-glass has good dielectric properties and is occasionally used in aircraft to minimize the impact of lighting strikes. E- and S-glass are used in the form of epoxy-based pre-preg or as fabrics containing unidirectional, woven, or chopped strand filaments.

A major advantage of E-glass fibers over the other types of fibers used in aircraft is their low cost. Figure 8.1 compares typical material costs for E-glass composites against costs for carbon, aramid (trade name, Kevlar), and boron/epoxy composites; the relative cost of boron pre-preg shown is divided by a factor of 10 to make the chart readable. Costs are given for composites made of pre-preg or fabric (woven roving, chopped strand mat). The costs are approximate and do not include the expense of fabricating the composite into an aircraft component, which is usually much higher than the raw material cost. E-glass composites are

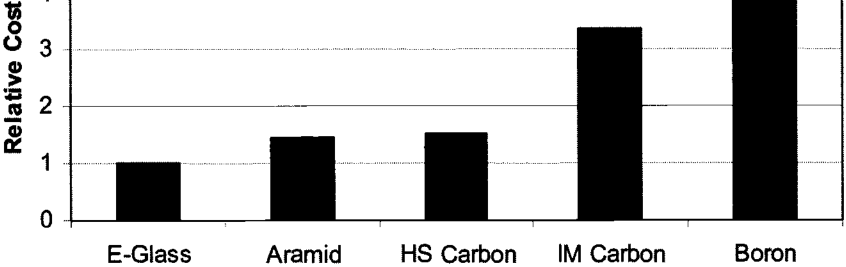


Fig 8.1 Relative costs of some fiber composite systems used in aerospace applications. Boron is shown at about 1/10 of its actual relative cost.

by far the cheapest, particularly when chopped strand mat or woven fabric is used. S-glass composites are much more expensive than the E-glass composites and only marginally less expensive than carbon/epoxy.

Figures 8.2 and 8.3 provide comparisons[1] of the strength and stiffness of some of the available forms of E-glass fiber materials. The forms are chopped-stand mat, woven rovings, and unidirectional pre-preg material. The comparisons in these figures are based on relativities that will also be relevant to the other fiber types if made from similar geometrical forms.

Table 8.1 provides relevant physical, thermal, and mechanical property data for unidirectional E- and S-glass/epoxy composites. Glass fibers have a specific gravity of about 2.5 g cm^{-3}, which is slightly lower than the density of boron fibers (2.6 g cm^{-3}) but is appreciably higher than carbon ($\sim$1.8 g cm^{-3}) and Kevlar (1.45 g cm^{-3}) fibers. The specific gravity of thermoset resins is around 1.3 g cm^{-3}, and as a result, glass/epoxy composites have a specific gravity that is higher than for other types of aerospace composites (except boron/epoxy) with the same fiber volume content. However, depending on the fiber volume fraction, it is still somewhat lower than that of aircraft-grade aluminum alloys (2.8 g cm^{-3}). The Young's moduli and strengths of both E- and S-glass composites are lower than those of other aerospace structural composites and metals. The combined effects of low stiffness and high specific gravity makes glass/epoxy or

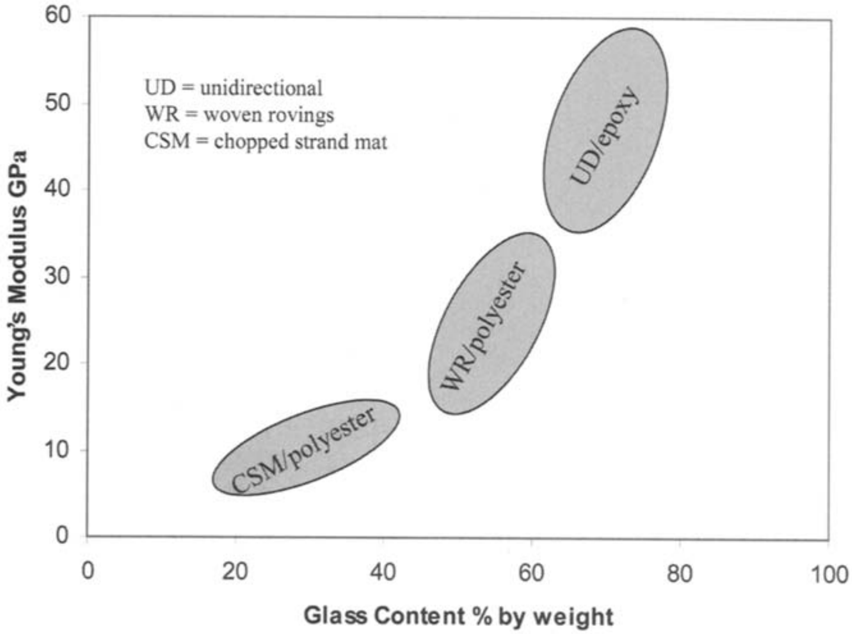


Fig 8.2 Typical Young's modulus for various types of glass-fiber composites. Adapted from Ref. 1.

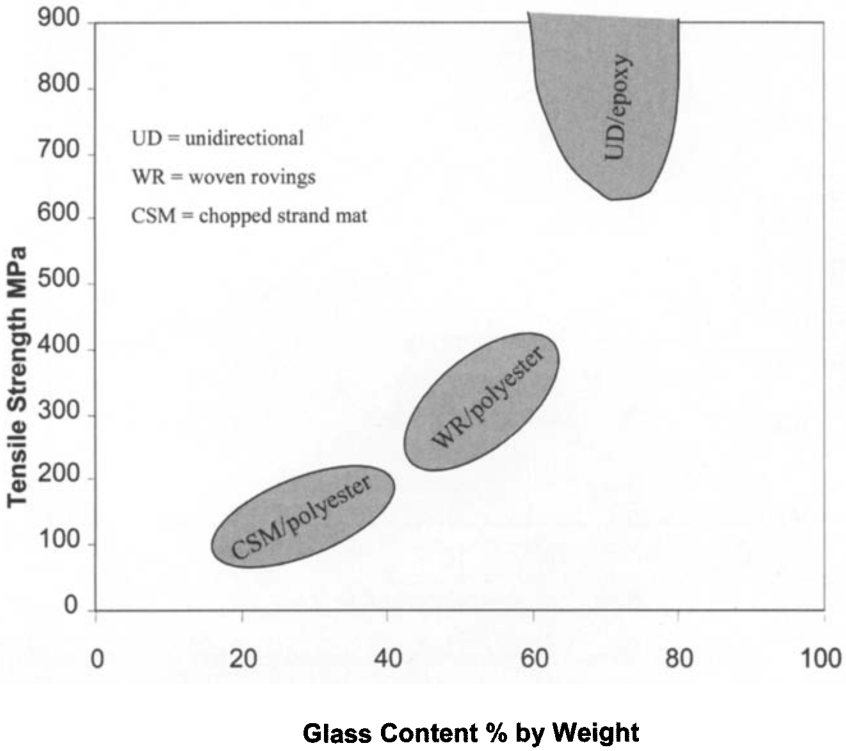


Fig 8.3 Typical strengths of various types of glass-fiber composites. Adapted from Ref. 1.

other glass fiber composites unattractive for use in weight-critical load-bearing primary structures on larger aircraft.

8.2.1 Fatigue Performance of Glass-Fiber Systems

Another drawback of using glass/epoxy composites in aircraft structures is their relatively poor fatigue performance compared with the other composites discussed in this chapter. Glass/epoxy composites are more prone to fatigue-induced damage (e.g., microscopic cracks, delaminations) and failure than other aerospace composite materials. Figure 8.4 shows a typical fatigue-life curve for a unidirectional glass/epoxy composite that was tested under cyclic tension-tension loading. Fatigue-life curves for unidirectional carbon/epoxy and Kevlar/epoxy laminates that were also tested under tension-tension loading are shown for comparison. In the figure, the normalized fatigue strain ($\varepsilon_f/\varepsilon_o$) is the maximum applied cyclic tensile strain (ε_f) divided by the static tensile failure strain of the composite (ε_o). Of the three materials, the fatigue-life curve for the glass/epoxy

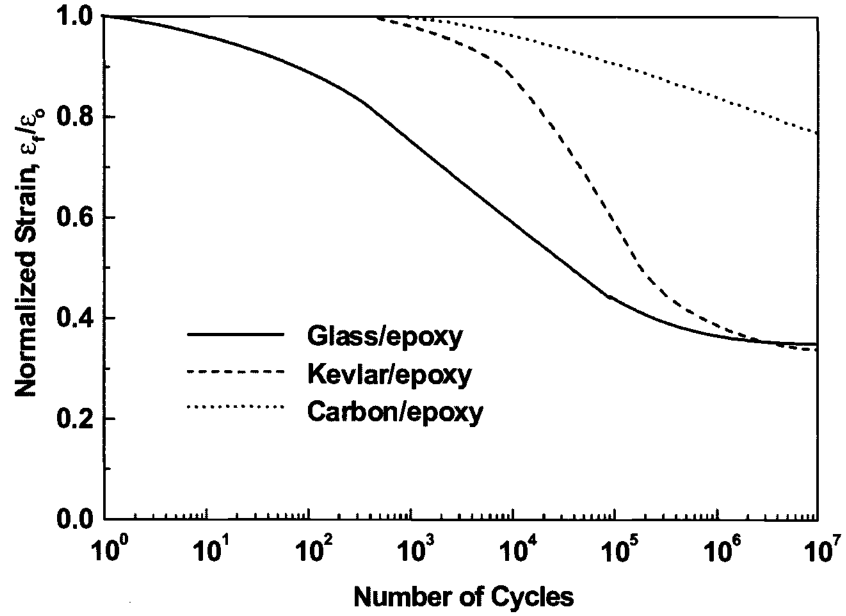


Fig 8.4 Fatigue-life curves for unidirectional composites subject to tension-tension loading.

composite drops the most rapidly, with increasing number of cycles. This indicates that glass/epoxy is the most susceptible to fatigue-induced failure under tension-tension loading, and this is due to the low stiffness of the glass reinforcement, resulting in damaging strains in the matrix, as discussed in Section 8.8.

The fatigue performance of glass/epoxy composites is degraded further when cyclic loading occurs in a hostile environment, such as in hot and wet conditions. The microscopic cracks and delaminations caused by fatigue loading create pathways for the rapid ingress of moisture into the composite. Moisture can then cause stress-corrosion damage to the glass fibers, which may dramatically reduce the fatigue life. Cracks and delaminations caused by fatigue also cause large reductions to the stiffness and strength of glass/epoxy. Figure 8.5 shows that the static tensile modulus and strength of [0/90]s glass/epoxy composites decrease rapidly with increasing number of load cycles before reaching a constant level. The residual modulus and strength remain relatively constant until near the end of the fatigue life. For some glass/epoxy materials, a reduction in stiffness and strength can occur within the early stage of the fatigue process, when the damage is not visible.

Finally, glass fibers composites and other composites having fibers with low thermal conductivity and low stiffness are prone to heat damage under cyclic

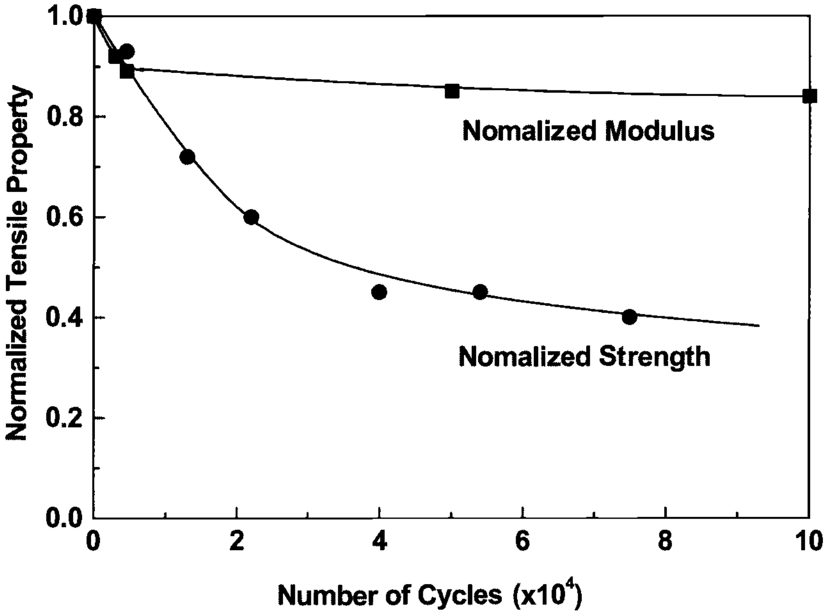


Fig 8.5 Effect of number of tensile load cycles on the Young's modulus and strength of a $[0/90]_S$ glass/epoxy composite.

loading at high frequencies[2] above around 5 Hz. This is because the heat generated by stress/strain hysteresis in the polymer matrix cannot be easily dissipated. The problem increases in thick composites, in which heat dissipation is even more difficult and with off-angle fibers where matrix strains are higher.

The performance of composites under cyclic loading is discussed further in Section 8.

8.2.2 Impact Strength of Glass-Fiber Systems

Although many mechanical and fatigue properties of glass/epoxy composites are lower than those of other carbon/epoxy and aramid/epoxy materials, they generally have a superior ability to absorb energy during impact. Figure 8.6 illustrates the relative energies for failure under impact of glass fiber and other composites considered in this chapter and some aluminum alloys, as measured with the Charpy test method. The exceptionally high impact toughness of S-glass fibers has led to their use in ballistic protective materials.

As shown, glass/epoxy composites have the highest impact energies, with S-glass/epoxy composites being 4–7 times more impact-resistant than high-strength carbon/epoxy laminates and about 35 times more resistant than high-modulus carbon/epoxy materials. Glass/epoxy composites are even 9–11 times more impact-resistant on this basis than aircraft-grade aluminum alloy.

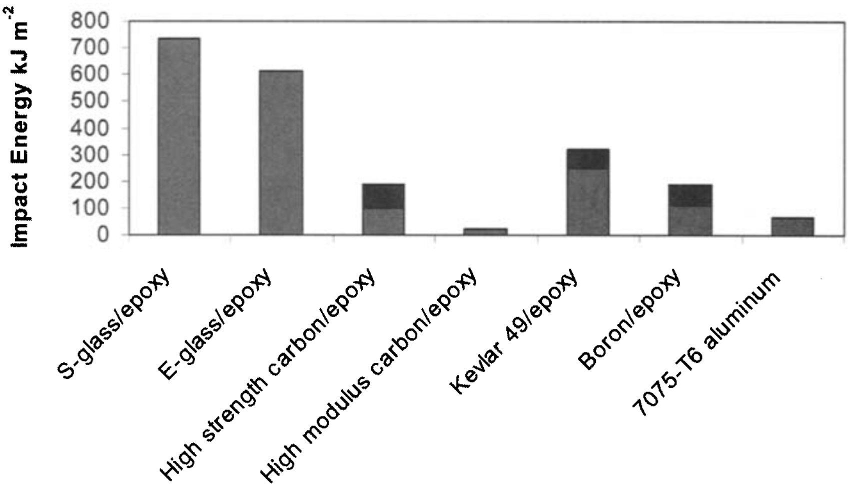


Fig 8.6 Charpy impact energy absorption of some composite and, for comparison, non-composite materials, as indicated.

8.2.3 Stress and Environmental Effects

As discussed in Chapter 3, glass fibers are prone to fracture when subjected to high stress for prolonged periods of time. This behavior, known as static fatigue or stress rupture, is exacerbated by exposure to moisture, as shown in Figure 8.7.

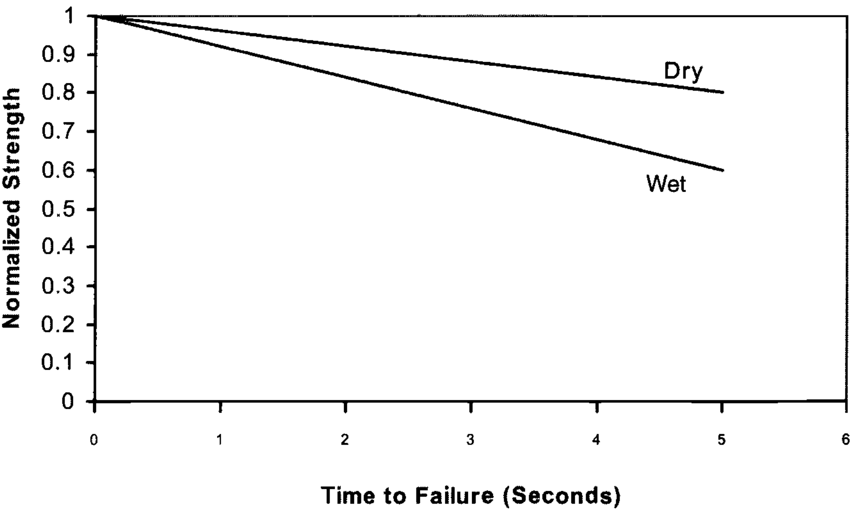


Fig 8.7 Stress rupture strength of E-glass fibers in air and water. Adapted from Ref. 1.

The influence of moisture on fiber strength is much reduced if the fiber is embedded in a polymer matrix, but can still be of concern in highly loaded applications such as pressure vessels.

Glass fiber composites, when exposed to moist environments or other aggressive environments, are also prone to degradation caused by weakening of the fiber/matrix interface. This weakening generally occurs by chemical attack at the fiber surface. The degree of weakening experienced depends on the matrix, the coating (called size or finish) used on the fiber, and the type of fiber. Weakening of the interface will result in significant loss in matrix-dominated mechanical properties such as shear, off-angle, and compression strength.

Environmental degradation is thus of significant concern for structural applications in which the ability to carry high loads is required and particularly where the loading is sustained.

8.3 Boron Fiber Composite Systems

Boron fibers (Chapter 3) were first discovered in 1959 and were subsequently developed during the 1960s into the first true high-performance fibers. Until that time, glass fiber was the only other high-strength fiber available in continuous lengths, and the low modulus of glass severely restricted its use in high-performance structures. The high-temperature capability of boron also provided the opportunity for producing metal-matrix composites, although it was a boron/epoxy (b/ep) composite that produced much of the initial commercial success. These composites were used successfully in several important aircraft component programs during the 1970s including the skins of the horizontal stabilizers on the F-14 and the horizontal and vertical stabilizers and rudders on the F-15. Boron/epoxy pre-preg materials are currently available in commercial quantities, and their unique properties make them suited to a range of specialized applications.

Because of the presence of a dense tungsten boride core (Chapter 3), the diameter of boron fibers is significantly greater than that of carbon fibers, to minimize fiber density and to ensure the properties of the fiber are not greatly influenced by the properties of the core. Fibers are currently produced in 100- and 140-μm diameters and therefore boron fibers have a very high bending stiffness (proportional to the fourth power of the radius). This restricts the radius that the composite can be formed into. For the 100-μm diameter fibers, a radius of around 30 mm is the practical limit. Although this is not of concern in the production of large, relatively flat aircraft components, it is sometimes a limiting factor in the selection of this composite system for the manufacture of a part with complex geometry.

The large diameter of boron fibers means that it is virtually impossible to weave these fibers into a fabric in the same way that glass, kevlar, and carbon fibers can. It is, however, possible to hold parallel boron fibers together with a weft thread of polyester to form a dry unidirectional preform. Boron pre-pregs are unidirectional and have a fine polyester scrim material (similar to that in structural film adhesives) incorporated into the resin on one side of the fibers to provide some lateral strength to the pre-preg during handling.

8.3.1 Mechanical Properties of Boron-Fiber Systems

Typical properties of unidirectional boron/epoxy composites are shown in Table 8.1. Boron composites typically have high compressive strength due to the large-fiber diameter, and this is one of their distinguishing features compared with carbon composites. Most of the advanced composite systems provide a significant improvement in specific stiffness over the conventional aircraft metallic materials, which have a common specific stiffness of around 25 GPa. Also apparent from Table 8.1 is the fact that although the density of cured boron composites is higher than carbon composites, it is appreciably lower than that of aluminum or titanium.

There are several types of carbon fibers on the market, some of which have properties that the densites of exceed either the tensile modulus or strength of boron fibers. Boron fiber composites, however, still have a blend of tensile and compressive properties that no single carbon fiber type is able to match. A form of pre-preg is available in which boron and carbon fibers are mixed together in the same pre-preg and this is marketed by Textron Specialty Materials as Hy-Bor. The properties of this material exceed those of conventional boron/epoxy composite due to the higher volume fraction of fibers.

8.3.2 Handling and Processing Properties of Boron-Fiber Systems

Boron is an extremely hard material with a Knoop value of 3200, which is harder than tungsten carbide and titanium nitride (1800–1880) and second only to diamond (7000). Cured boron composites can be cut, drilled, and machined with diamond-tipped tools, and the pre-pregs are readily cut with conventional steel knives. In practice, the knives cannot actually cut the hard fibers; however, gentle pressure fractures the fibers with one or two passes. Although it is possible to cut complex shapes with the use of templates, laser-cutting has been shown to be the most efficient way to cut a large amount of non-rectangular boron plies.

Boron fibers are currently available in several forms. As well as the two fiber diameters, pre-pregs are available with either 120°C or 175°C curing epoxies. With the exception of the reduction in formability mentioned above, in most other aspects, boron pre-pregs handle and process in a similar fashion to the more common carbon pre-preg materials.

8.3.3 Aircraft Applications of Boron-Fiber Composites

The fiber manufacturing process described in Chapter 3 shows that the fibers are produced as monofilaments on an expensive precursor filament, and this basic method has not changed since the early 1960s. This is the main reason that boron fibers are more costly than carbon fibers (an equivalent quantity of boron/epoxy pre-preg is roughly 12 times the price of carbon/epoxy pre-preg). The high cost of boron fiber was, initially, not critically important in defense applications and, because of its excellent specific mechanical properties, was selected for some of the empennage skins in the F-14 and F-15, and is also used in the B-1 bomber, in several components. However, in the 1970s, as the quantity of carbon fiber production rapidly increased, the cost of carbon fibers fell considerably, so that for most common aircraft applications, it became a more cost-effective fiber than boron in other than specialized applications.

One application for which boron/epoxy is well suited is as a repair material for defective metallic structures.[3] When repairs to aircraft components are considered, for example, the amount of boron/epoxy required is usually not great, and so the comparatively high cost of the material is not a critical factor. The high specific tensile and compressive properties of b/ep are ideally suited to repair applications. Carbon/epoxy can also be used for these applications; however, this material has several disadvantages. Because repairs are adhesively bonded to the structure with high-temperature curing adhesives, the lower coefficient of expansion of carbon/epoxy results in higher residual stresses in the repaired structure. These residual stresses can increase the local stresses at the defect. In addition, carbon fibers are electrically conducting, which inhibits the use of eddy-current non-destructive inspection methods through the repair material to confirm that there has been no growth of the damage. Boron fibers do not produce a galvanic couple with aluminum, so there is no danger of a boron repair causing corrosion of an aluminum aircraft structure.

8.4 Aramid Fiber Composite Systems

When Kevlar 49/epoxy composites were introduced by DuPont in the mid 1960s, they had a higher specific tensile strength than similar composites, based on the then available carbon fibers. However, the subsequent development of carbon fibers with greatly improved strength properties displaced aramid composites from this position. Now they fill a property gap in specific strength and stiffness between glass and carbon fibers.[4]

In contrast to their high tensile properties, compression strength of aramid composites is low. Under compression loading, aramid fibers[5] undergo non-linear deformation at strain levels around 0.5% by the formation of kink bands. Essentially, this mode of deformation occurs because the extended chain structure of the aramid fibers is unstable under compression loading. Figure 8.8 illustrates the extreme asymmetry in stress/strain behavior tension and compression loading for a typical aramid/epoxy composite.

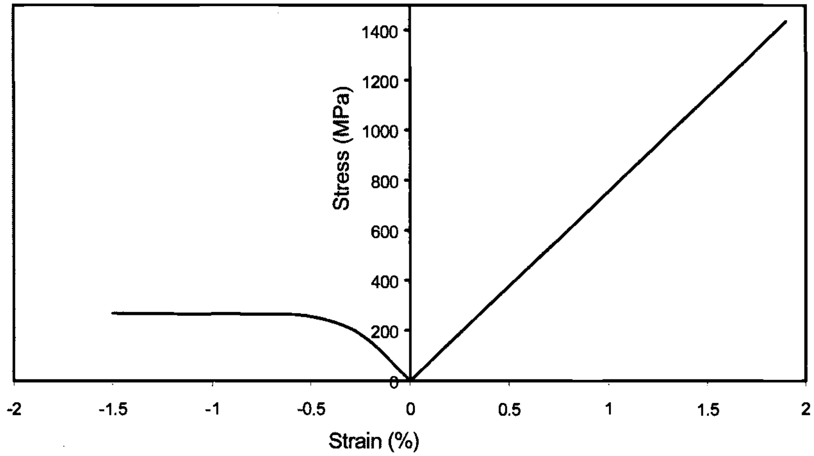


Fig 8.8 Typical tensile and compression stress-strain curves for aramid composites at ambient temperature. Adapted from Ref. 4.

The low compression resistance of aramid composites is a major disadvantage in applications requiring high strength or stiffness under compression or flexural loading. However, the non-linear behavior in compression, combined with a high strain capacity under tension, is a significant advantage in applications in which resistance to severe mechanical contact or penetration is required. Thus, in aerospace applications, aramid composites were favored for use in secondary structures such as fairings subject to impact damage. Thin-skin honeycomb panels based on aramid fibers were used extensively in some civil applications; however, the skins suffered from severe moisture penetration. This problem was mainly attributed to microcracking of the skins, possibly caused in part by moisture absorption and swelling of the fibers, coupled with the relatively weak fiber-to-resin bond strength.

The properties of high tensile strength and resistance to penetration damage continue to favor aramid composites for use in filament-wound vessels and for containment rings in engines. Ballistic protection is another important use of aramid composites, for example, in structural or non-structural components on helicopters for protection against small arms fire.

Finally, aramid fibers are used as the reinforcement in aircraft radomes, as they have favorable dielectric properties.

For components that require both good compressive properties and impact resistance, aramid fibers may be used in combination with carbon or glass fibers. They can be used to enhance the toughness properties of carbon-fiber composites or to improve strength in the presence of stress raisers. Hybrid aramid/carbon composites have been used in helicopter fuselage panels and in civil aircraft for fairings.

8.4.1 Manufacturing Issues with Aramid Composites

A significant issue in manufacturing aramid-fiber composites is the difficulty in achieving adhesion between the fibers and polymer matrix. Thus, the fibers must be surface-treated to enhance adhesion. However, in some applications, notably those requiring good ballistic properties, a fairly low-level of adhesive strength between fiber and matrix is desirable to obtain optimum energy absorption properties. In the case of aramid filament-wound pressure vessels, burst strength is highest at some intermediate level of bond strength.

Various treatments have been used to improve fiber/matrix adhesion,[6] including gas plasma treatment in Ar, N_2, or CO_2, which typically results in a 20% improvement in interfacial bond strength to epoxy.

Aramid fibers absorb moisture, up to around 6% by weight, if exposed to a humid environment. This can affect fiber/matrix adhesion and other properties, so the fibers are either stored in low humidity conditions or dried before usage.

8.4.1.1 Matrix Systems for Aramid Composites. Some thermoset resin systems such as anhydride-cured bisphenol A epoxies are inherently more compatible with aramid fibers than other matrix resins and provide relatively high interlaminar strengths. Vinyl esters are more compatible with aramid fiber than polyesters and are used for marine-type applications. To obtain optimum tensile properties, it is important that the resin has high elongation. About 6% appears to provide the best balance of properties. Thermoplastic such as PEEK and polysulphones can also be successfully used. However, as processing temperatures can exceed 260°C in the case of polysulphones, and as high as 400°C in case of PEEK, there is some degradation of the fiber strength.

8.4.1.2 Cutting, Drilling and Machining Aramid Composites. The high toughness of aramid fibers, including their tendency to defibrillate (separate into microfilaments) under high compressive and shear stresses, makes aramid composites very difficult to cut or machine. Indeed, dry aramid cloths themselves are difficult to cut and require the use of special shears, although heavy-duty upholstery scissors can be used. Special carbide-tipped tools are required for drilling and machining. Water jet is an excellent method for cutting aramid composites and also minimizes the creation of airborne fibers.

8.4.2 Mechanical Properties of Aramid Composites

As mentioned previously, under tensile loading, the strength of aramid/epoxy pre-preg laminates can match or exceed those of similar carbon/epoxy or glass/epoxy composites. Their elastic modulus is below that of carbon/epoxy but exceeds that of glass/epoxy. Typical values, including those for similar carbon and E-glass/epoxy composites, are listed in Table 8.1 The Table shows that although the elastic modulus is similar under tension and compression loading, strength is much reduced. Apparent interlaminar shear strength (ILSS) is also relatively low compared with the other composites. One reason for this is the low fiber/matrix bond strength. Another reason is the poor compression properties of these composites because failure in the standard short-beam ILSS test is rarely pure shear and often includes a significant component of compression failure.

8.4.2.1 Fatigue Resistance. Under tensile-dominated cyclic loading, as illustrated schematically in Figure 8.9, unidirectional aramid composites are superior to aluminum alloys and to S-glass/epoxy composites but inferior to carbon/epoxy (not shown). For unidirectional composites, the fatigue damage occurs mainly as matrix microcracking. As may be expected, the rate of damage accumulation depends on the strain level experienced by the matrix, which is directly dependent on the fiber elastic modulus and volume fraction—hence, the relative ranking. The relative advantage of the composites over aluminum alloys is reduced in cross-plied laminates, normally used in aircraft structures. Nevertheless, a marked advantage over aluminum alloys is maintained for the aramid- and carbon-fiber composites.

As is to be expected from the poor compression strength of the fibers, aramid composites are inferior to both glass and carbon composites under compression-dominated fatigue.

8.4.2.2 Creep and Stress Rupture. Aramid fibers and composites have a similar low creep rate to glass fibers but, as illustrated in Figure 8.10, they are less

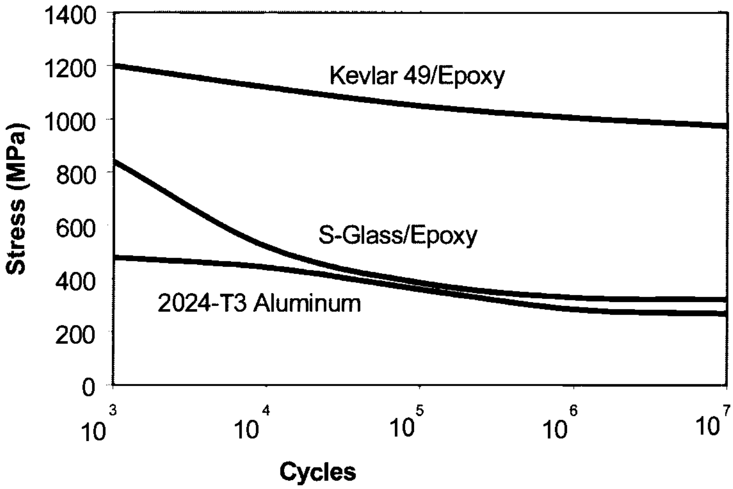


Fig 8.9 Plot of tension-tension fatigue results for unidirectional composites and for an aluminum alloy. Adapted from Ref. 4.

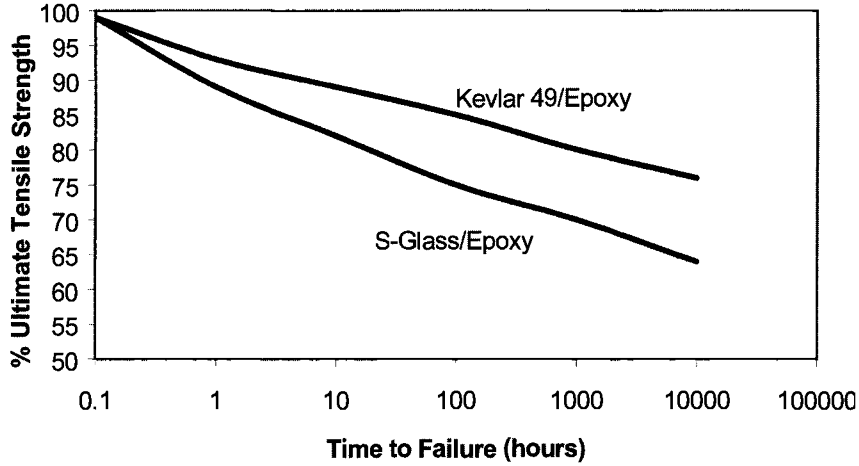


Fig 8.10 Stress rupture properties of unidirectional aramid and glass fibers in epoxy resin. Adapted from Ref. 4.

prone to stress rupture. Glass fibers are particularly sensitive to humid environments, where they have much lower stress rupture properties. Generally, carbon fibers are significantly more resistant to creep and stress rupture than glass or aramid fibers. Although, in unidirectional composites, the creep behavior is dominated by the fiber properties, the relaxation of the matrix makes a small contribution to the relatively short-term creep behavior. The creep rate increases and the stress rupture decreases as a function of both temperature and humidity.

8.4.2.3 Environmental Effects. Aramid fibers absorb moisture; at 60% relative humidity, the equilibrium moisture content is about 4%, which rises to around 6% when the RH is 100%. The result is a decrease of tensile strength and stiffness at room temperature of around 5% (probably significantly greater at elevated temperature), which would be reflected in the properties of the composite. However, the effect of moisture on the fibers appears to be reversible. Tensile strength of the dry fiber is reduced by up to 20% at 180°C. Room temperature strength is also reduced by about 20% after prolonged (80 h) exposure at 200°C.

The effects of temperature and moisture on tensile and compression properties are illustrated in Figure 8.11. Tensile properties are unaffected up to a relatively high temperature (177°C) when the loss is around 30% hot/wet. The loss in compression strength at this temperature is quite dramatic and is around 70%. However, similar carbon/epoxy composites would also experience a significant loss of compression strength under wet conditions close to the cure temperature.

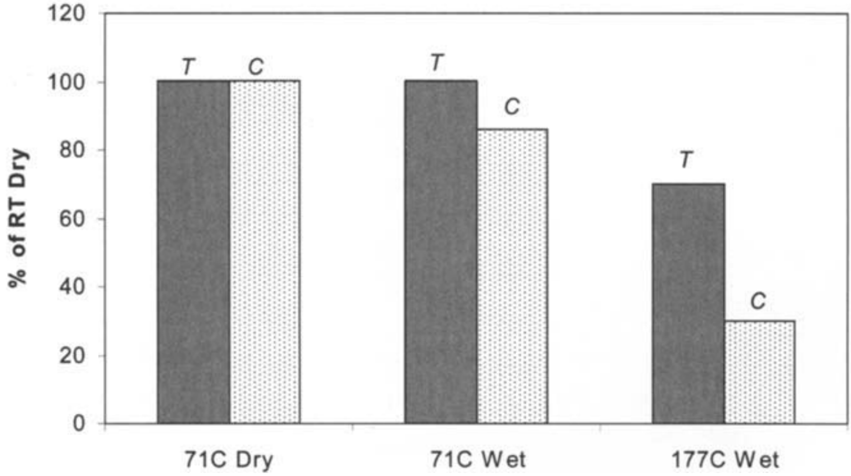


Fig 8.11 Effect of temperature and moisture on tensile T and compression strength C of Kevlar 49/epoxy composites in a 171 °C cure epoxy resin, compared with value at room temperature. Adapted from Ref. 4.

8.4.3 Other Useful Properties of Aramid Composites

8.4.3.1 Impact and Ballistic Properties. Aramid composites have the capacity to absorb large amounts of energy during penetration (Figure 8.12). In part, this is due to the high strain-to-failure and moderate elastic modulus that

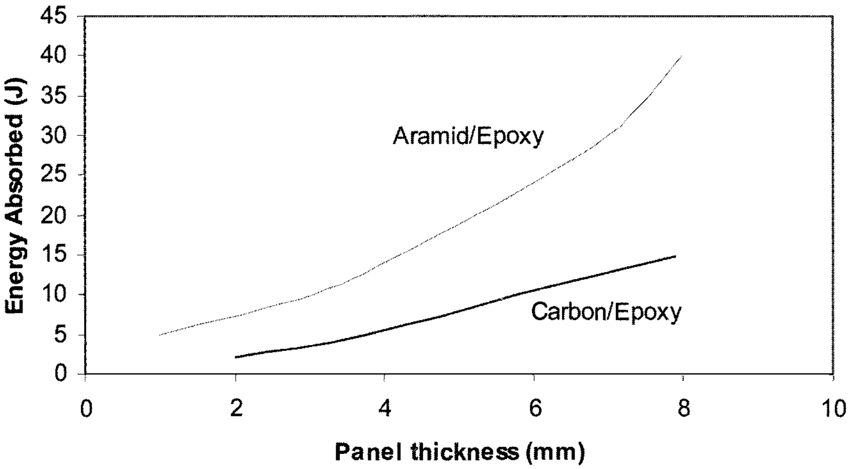


Fig 8.12 Drop-weight impact resistance of aramid/epoxy (Kevlar 49) and carbon/epoxy (Thornel 300) quasi isotropic laminates in Hexcell F-155 resin. The energy parameter is for through-cracking, but not penetration. Adapted from Ref. 4.

results in a very large area under the stress-strain curve. This is an indication of the large energy-absorbing capacity of aramid composites in tension. In addition, the complex fiber-failure modes involving kinking in compression and defibrillization during final fracture, together with the strong tendency for disbonding, greatly add to the energy-absorbing capacity of aramid composites under dynamic loading.

One way of comparing ballistic performance of composite laminates is based on the V_{50} parameter. This is the velocity at which there is a 50% probability that the projectile will penetrate a target of the laminate. The V_{50} number for laminates of a given areal density is one way of making the comparison, where the areal density is the thickness multiplied by the density. Alternatively, the areal density for a given V_{50} can be used as the basis of comparison. Figure 8.13 shows results for a Kevlar (aramid) composite compared with S- and E-glass composites. This shows that S-glass composites provide a level of protection similar to that of aramid, and that both are much superior to E-glass and an aluminum alloy.

8.4.3.2 Vibration Damping. Composites based on aramid fibers exhibit very high damping qualities, particularly under reversed cyclic loading. In part, the high damping results from the non-linear deformation of the fibers in compression. This is an important advantage of aramid-fiber composites for aircraft applications where reduced noise and vibration are design objectives. Figure 8.14 illustrates the damping behavior of aramid composites compared with some other structural materials also having relatively high damping.

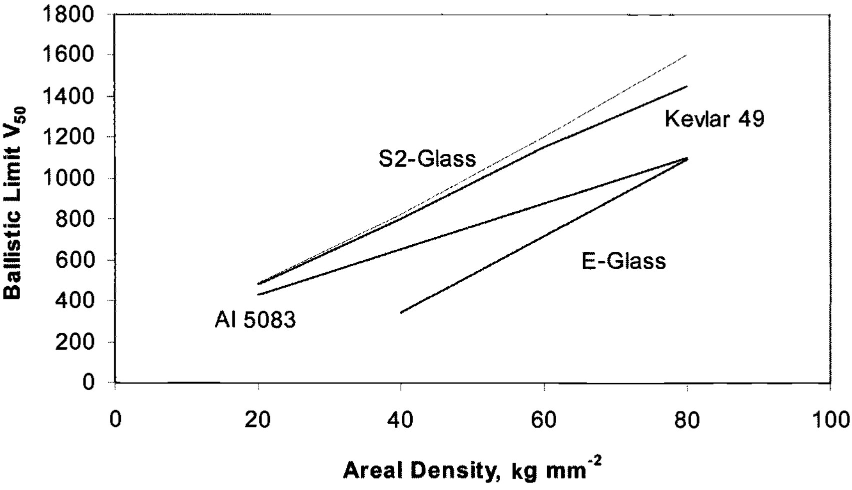


Fig 8.13 Relative ballistic performance of lightweight armor materials. Adapted from Ref. 4.

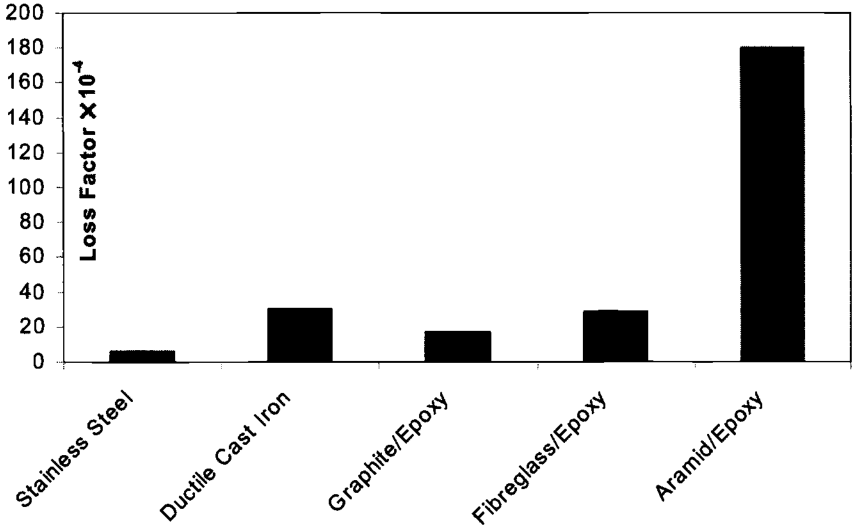


Fig 8.14 Loss factor from decay of free vibration for various materials. Adapted from Ref. 4.

8.4.3.3 Aramid Composites for Pressure Vessels and Containment rings. Aramid composites are particularly well suited for use as pressure vessels because of their excellent specific tensile properties and their resistance to mechanical damage. The comparative performance of pressure vessels is often made on the basis of the parameter PV/W where PV is pressure $\times$ volume and W is the weight. A comparison on this basis of pressure vessels made with the three fiber types in epoxy matrices is shown in Figure 8.15. The influence of a 20-J impact on strength, adapted from some relevant data, is also shown.

As a result of their excellent performance under pressure loading and their damage resistance, aramid composites are frequently used as containment rings for jet engines, which prevent fractured engine parts (such as broken blades), exiting the casing of the engine, and damaging other parts of the aircraft.

8.4.3.4 Properties of Aramid-Hybrid Composites. As discussed earlier, hybridization with fairly low volume fractions of aramid composites can be used to reduce stress concentrations around holes or cut-outs or to improve resistance to impact damage in carbon-fiber-based composites and to improve stiffness in glass/epoxy composites. Alternatively, carbon-fiber composites can be hybridized with aramid to improve toughness while maintaining the favorable compression strength properties of the carbon-fiber composites.

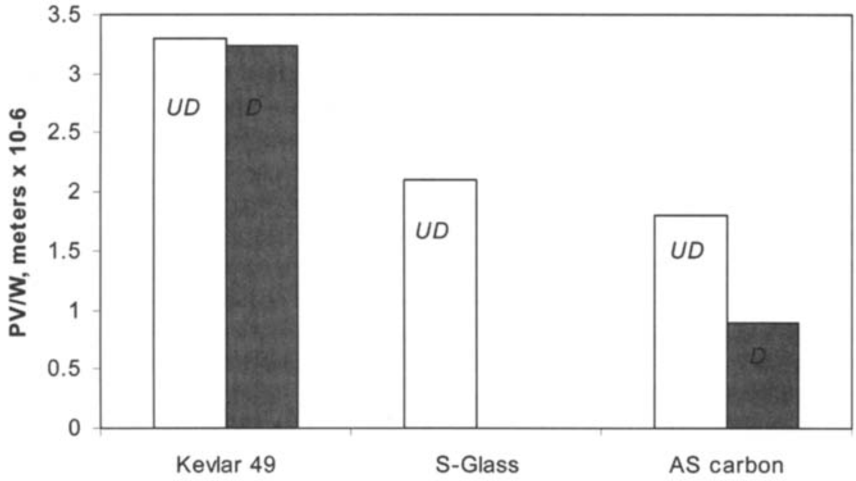


Fig 8.15 Performance of damaged (D) and undamaged (UD) pressure vessels made in the various composites. Adapted from Ref. 4.

Figure 8.16 shows the effectiveness of using 20% of 0° aramid layers in a quasi isotropic carbon/epoxy composite to improve open-hole tensile strength. The high-strain aramid fibers inhibit propagation of fiber and matrix cracking.

8.5 Carbon Fiber Systems

A discussion of the key mechanical properties of carbon-fiber composites is provided in Chapter 12 on design issues and in Chapter 13 on airworthiness issues, which should be read in conjunction with this chapter. The topic of impact damage of these composites is also covered in Chapter 12.

Carbon-fiber composite systems are used more extensively for structural applications within the aerospace industry than other high-performance fiber systems. This is primarily due to the overall high specific stiffness and strength properties that can be achieved from these composites compared with other composites and structural metals, as shown in Table 8.1.

The mechanical properties of carbon-fiber composites can be varied significantly through the choice of the carbon fiber. Table 3.1 in Chapter 3 lists the properties of the various grades of carbon fiber and Table 8.2 provides details of relevant mechanical properties of some carbon/epoxy systems widely exploited in aircraft structures.

PAN-based carbon-fiber composites dominate the market because of their lower cost, better handling characteristics (due to the higher failure strains of the

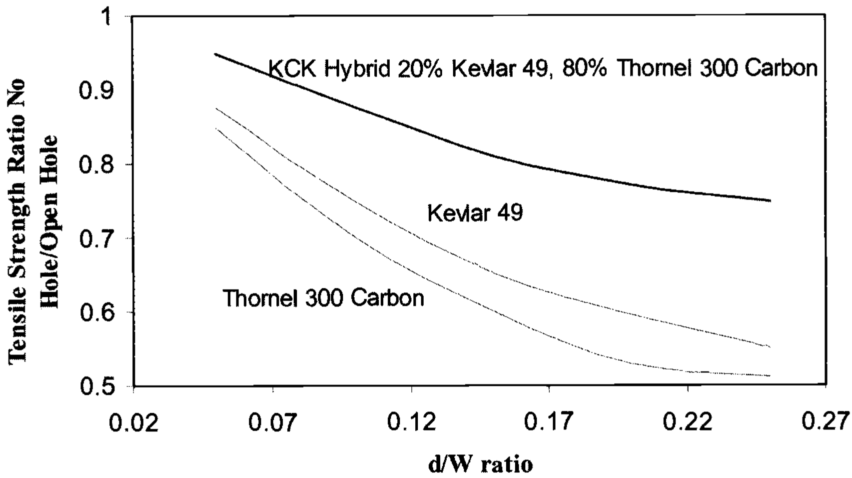


Fig 8.16 Tensile strength of laminates with open holes. Adapted from Ref. 4.

PAN fibers), and attractive overall composite mechanical properties. Pitch-based fiber composites tend to be used extensively in satellite applications, where their superior stiffness and thermal properties, including high conductivity and low coefficient of expansion, are a major advantage.

Carbon-fiber reinforcements can be produced in an extensive range of forms that also influence the properties of the composite system. These forms, ranging from short chopped fiber mats, through a variety of woven fabric products, to unidirectional tapes and advanced multi-layer fabrics such as non-crimp materials, are discussed in Chapter 3. Advanced forms such as three-dimensional carbon composites are covered in Chapter 14.

8.5.1 Matrix Systems for Carbon-Fiber Systems

The polymer matrix systems used with carbon fibers are discussed in Chapter 4. Thermoplastic matrices, such as PEEK and PPS, are becoming increasingly used for applications in the epoxy temperature range because of their higher toughness and low moisture absorption. However, recent developments in toughened epoxies have reduced the toughness advantage.

Epoxy resin systems have an upper limit of service temperature of around 180–200°C. For higher temperatures (up to approximately 250°C) BMI systems are often used, whereas for even higher temperatures (up to around 320°C) polyimide matrix systems may be used. Compared with epoxies, all the other matrix systems are more expensive and give rise to either processing or durability problems.

Table 8.2 Mechanical Properties of Selected Current Carbon-Fiber Composite Systems, Based on Data Provided in Ref. 2

Property	Units	Standard carbon/ epoxy AS4/3501-6	Intermediate-modulus carbon/ epoxy IM6/1081	High-modulus carbon/epoxy GY-70/934	Carbon/ thermoplastic AS4/PEEK	Standard carbon/ bismaleimide T300/V378
SG	—	1.58	1.6	1.59	1.57	1.6
V_f	—	0.63	0.65	0.57	0.58	0.65
σ_{1tu}	MPa	2280	2860	589	2060	1586
σ_{1cu}	MPa	1440	1875	491	1080	1324
σ_{2tu}	MPa	57	49	29.4	78	—
τ_u	MPa	71	83	49.1	157	—
E_1	GPa	142	177	294	131	138.6
E_2	GPa	10.3	10.8	6.4	8.7	—
G_{12}	GPa	7.2	7.6	4.9	5	—
ν_{12}	—	0.27	0.27	0.23	0.28	—
ε_{1u}	—	0.015	0.016	0.002	0.016	—
ε_{2u}	—	0.006	0.005	0.005	0.009	—

8.5.2 Adhesion and Bonding of Carbon Fibers in Composites

Carbon fibers are normally surface-treated to develop adequate bonding to epoxy or other matrices. To achieve adequate toughness, it is important that the fiber is able to disbond from the matrix to alleviate local stress concentrations, for example, at matrix cracks. Ductile matrices such as those based on thermoplastic matrices may inhibit disbonding, possibly resulting in reduced toughness and fatigue properties.

Surface treatments are based on oxidation of the fiber surface either by a wet chemical process (e.g., with sodium hypochlorite or chromic acid) or a dry process involving ozone. The treatment removes weak films, roughens the surface on a microscopic scale, and introduces chemically active sites onto the fiber. Only minor weakening of the fibers occurs as a result of these treatments. A coating of a compatible resin, generally similar to the thermosetting matrix resin, is sometimes used to protect carbon fibers from damage during reinforcement manufacture and also to provide lubrication. The coating can also improve adhesion and wetting.

There are various methods of measuring the fiber/matrix bond strength, including bulk tests based on measurement of transverse strength of composites and single-fiber tests.[7]

It appears that an upper value for the shear strength of the fiber/matrix bond for standard carbon fibers such as AS4 in an epoxy matrix is between 40–75 MPa, depending on the strength and ductility of the resin system. Interlaminar shear (ILSS) values for similar composites lie between 90 and 130 MPa. However, ILS tests often result in failure modes other than shear as the stress state in the failure zone is complex.

The fiber/matrix interface is often considered and modelled as the third phase in a composite, called the interphase. Significant effort has been directed at quantifying the effect of this phase on composite behavior in the case of carbon fiber composites.[8]

8.5.3 Effect of the Matrix and Fiber/Matrix Bond Strength of Carbon-Fiber Composites

8.5.3.1 Tension and Compression. Studies on unidirectional carbon-fiber composites made using a standard epoxy matrix and a range of fibers of differing properties surprisingly exhibit no consistent improvement in composite strain to failure with fibers of differing strain to failure or stiffness.[7] In general, however, high strain-to-failure matrices provide the best translation of fiber properties. Fiber surface treatment appears to have only a minor effect on tensile properties of the resulting composites.

Under compression loading, elastic properties of the matrix play a more important role because they support the fibers against microbuckling, which is the predominant failure mode. Also, as may be expected from this mode of failure, the straightness of the fibers is a very important factor contributing to good translation of fiber properties. Often, compression strengths are quoted at just 50–60% of the corresponding tensile value. However, compression strength of unidirectional composites is difficult to measure, as discussed in Chapter 7, so that different test methods can result in different conclusions.

8.5.3.2 Intra-and Interlaminar Properties of Carbon-Fiber Composites. It is to be expected that, for a particular carbon-fiber composite, intra- and interlaminar properties would depend strongly on the fiber/matrix bond strength, which is related to the level of surface treatment. In Ref. 7, it is shown that an improvement of around 100% in transverse strength (and interlaminar toughness, G_{Ic}) is obtained after applying a surface treatment of only 25% of that required to achieve maximum fiber/matrix bond strength.The values of the interlaminar tensile strength are nominally similar to the transverse tension strength. However, as mentioned earlier, ILSS values obtained from short-beam shear tests are often significantly higher, but direct comparison is not possible because this test produces complex loading and multiple failure modes.

Interlaminar toughness is related both to the properties of the matrix and the fiber/matrix bond strength. Further, the matrix is highly constrained by the fiber and therefore cannot achieve its potential toughness. This behavior is well known in adhesive bonding, where the toughness is greatly reduced when the adhesive thickness is not sufficient for full development of the plastic zone at the crack tip. Thus, a direct correlation of toughness of the composite with matrix toughness properties may not be expected, at least for the tougher matrices.

Similar comments can be made regarding mode 2 fracture toughness (which is significantly higher than mode 1 toughness). Interlaminar strength and impact resistance, discussed later, are expected to be related to interlaminar toughness. This is the case although the relationship is generally not straightforward and depends on mixed-mode (combined mode 1 and mode 2) behavior. Some of these issues are discussed in relation to adhesive bonds in Chapter 9.

8.5.3.3 Long-Term Deformation Behavior. Carbon fibers do not show any significant increases in strain with time (creep) over the working temperature range and are significantly less susceptible to stress rupture than aramid- or glass-fiber composites.[2] Thus, for a unidirectional composite under tensile or compressive loading, creep deformation will be low, and what does occur will result from loss of stiffness due to stress relaxation in the matrix. The situation regarding creep is, however, quite different for highly matrix-dominated composites, such as one based on $\pm 45°$ ply layers, where significant creep or stress-relaxation occurs at elevated temperatures. However, creep is not expected to be a major concern for a quasi isotropic laminate working within its stress and temperature design range.

8.6 Properties of Laminates

8.6.1 Tensile Strength of Cross-Ply Properties

Multilaminate cross-plied composites are, in the majority of cases, made up of families of 0°, ±45° and 90°, plies (although other ply angles are sometimes included, for example 15°, 30° and 60°).

As discussed in Chapter 2, very significant losses in strength in unidirectional materials and changes in failure mode occur when the load is aligned at small angles to the 0° direction. The resulting off-axis mechanical properties depend on both the matrix and fiber/matrix interface properties. When considering the tensile strength of cross-plied laminates, it is obvious that the strength of the laminate will be determined by the capacity of the stiffer 0° plies (providing that these are present in sufficient proportions) because these plies are the most highly loaded. However, the strain level that can be achieved by the 0° plies in the cross-plied laminate is usually much more dependent on the matrix and fiber/matrix bond strength properties than is the case for the unidirectional material and is also dependent on the specific ply configuration.

As may be expected, fracture behavior is even more complex under combined loading, particularly when some fibers are in compression. A further issue is that for composites with brittle matrix or low fiber/matrix bond strength, the off-axis plies (most usually the 90° plies) may have a lower strain capacity than the 0° plies and may crack first. Cracking before final failure of the 0° plies (and hence the laminate) is called first ply failure (FPF). FPF usually occurs in the form of fine microcracks and does not greatly affect stiffness of the laminate; however, it can greatly aid the penetration of the environment into the composite and can lead to delamination and a lower strain-to-failure of the 0° plies. In some cases, due to high residual stresses, microcracking may occur in the absence of external load; this was a notable feature of some of the very early carbon/epoxy composites because of the very low strain capability of the epoxy matrix.

As previously mentioned, some microcracking leading to minor delaminations is actually desirable at stress concentrators, such as holes, because they can markedly reduce stress concentrations.

Even with a brittle matrix system the adverse effect of 90° ply cracking on the 0° plies is much less in a thick laminate with the 0° and 90° plies concentrated into thicker layers, because delamination of these layers and therefore removal of the stress concentration is likely. Fine dispersion of 0°/90° plies is likely to have the opposite effect.

Factors other than those discussed above that can affect strength of cross-plied laminates include mode of loading, ply stacking sequence, presence of free edges, specimen width, and residual stresses, which in some cases can be high enough to cause failure, even in the absence of external stresses.

Chapter 6 describes the various approaches used to estimate the strength of cross-ply laminates under complex loading.

8.7 Impact Damage Resistance

Figure 8.6 provides a comparison of Charpy impact behavior of various composites. This shows that the S-glass-fiber composites have the highest capacity for energy absorption followed by E-glass and then aramid composites. High-strength carbon/epoxy has a significantly lower energy-absorbing capability than these materials and high-modulus carbon/epoxy, the lowest of all the composites.

The capacity for energy absorption of some glass and aramid composites greatly exceeds that of aluminum alloys and even steel. This behavior is attributed to the high-strain capacity of fibers when loaded in the fiber direction. However, other than providing some idea of energy absorption capacity under dynamic loading conditions such as in a crash, these tests provide no information on the important issue of the effect of impact damage on residual strength and stiffness—that is, the remaining strength of the composite structure after damage. Here, the concern is the effect of in-service impacts in the plane of the laminate.

Impact damage can result, for example, from dropped tools, runway stones, or large hailstones. The drastic reduction in residual compression strength and less reduction in tensile strength that can result from impact damage is a major issue in the design and airworthiness certification of these composites, as discussed in Chapters 12 and 13.

The type of damage resulting from impact on composites depends on the energy level involved in the impact. High-energy impact, such as ballistic damage, results in through-penetration with some minor local delaminations. Lower-energy-level impact, which does not produce penetration, may result in some local damage in the impact zone together with delaminations within the structure and fiber fracture on the back face. Internal delaminations with little, if any, visible surface damage may result from low-energy impact.

The actual damage response depends on many intrinsic and extrinsic factors, including the thickness of the laminate, the exact stacking sequence, the shape and kinetic energy of the impactor, and the degree to which the laminate is supported against bending. The strain-to-failure capability of the fibers will determine the degree of back-face damage in a given laminate, and the area of the damage depends on the toughness of the matrix and fiber/matrix bond strength as well as the failure strain and stiffness of the fibers. Also, composites based on woven fibers show less internal damage for a given impact energy than those based on unidirectional material. This is because damage growth between layers is constrained by the weave.

High and medium levels of impact energy thus cause surface damage that is relatively easily detected. Low-energy impact produces damage that is difficult to observe visually and is therefore commonly termed barely visible impact damage (BVID). This type of damage is of concern because it may occur at quite low-energy levels and is by definition difficult to detect.

Figure 8.17 shows the area of BVID as a function of matrix toughness. Residual strength with BVID correlates quite well with the area of the damage zone, although different composite systems will have somewhat different sensitivities, depending on the matrix toughness as well as other factors.

The morphology of a BVID level impact is shown in Figure 8.18, taken from Ref. 9. It is seen that the damage occurs within a conical (Hertzian) contact zone with the apex at the point of impact. Within the cone, the damage consists of delamination between and within plies, and on the back-face (base of the cone) fiber fracture. The fiber fracture in this region results from the high strain caused by local bending in a thin laminate. Thick laminates do not usually suffer back-face fiber damage; however, at high-impact energies, fibers are crushed at the point of impact.

Delaminations occur as lobes between plies of significantly different orientation (e.g., $+45°$, $-45°$, $0°$, $90°$) and extend in the direction of the outermost reinforcing ply. Within thick laminates, damage only occurs as interply cracking.

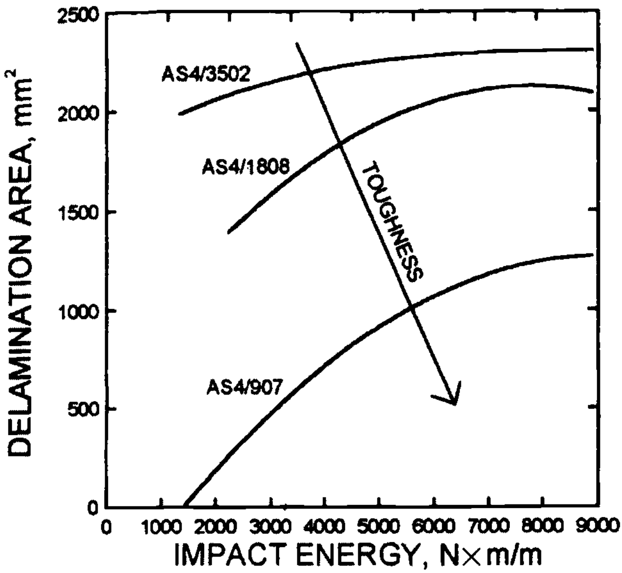
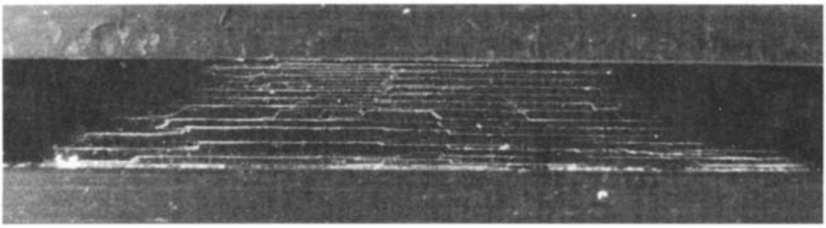


Fig 8.17 Delamination area as a function of impact energy for some carbon/epoxy composites with differing matrix toughness. Taken from Ref. 2.



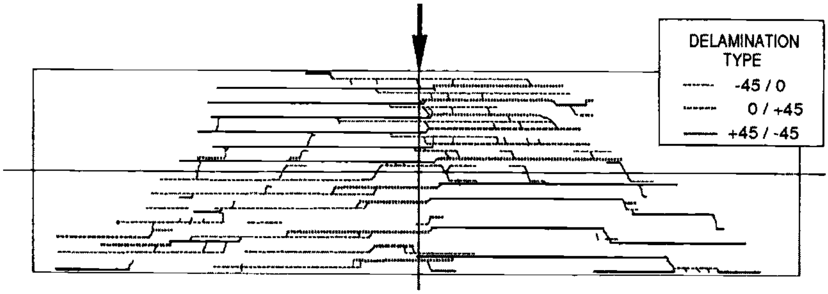


Fig 8.18 Impact damage in a 56-ply XAS-914C laminate and schematic of delamination pattern. Taken from Ref. 9.

8.7.1 Effect of BVID on Residual Strength

The effect of BVID on reducing residual compressive strength is well characterized experimentally. However, the actual mechanism has yet to be fully understood. It is clear that in the case of compression loading, the damage constitutes a zone of instability allowing the fibers to buckle at much lower strain levels than in the undamaged region. The marked effect of BVID on residual compression strength is shown in Figure 8.19 for four types of carbon composite.

In general, the reduction in residual strength is a similar function of damage size for all matrix systems. However, for a given impact energy, the damage size is less in tougher composite systems, such as those based on thermoplastic matrices.

Damage is most simply modelled as a softened zone or in more detail as a zone where the plies have become locally decoupled. Decoupling allows the plies to distort at relatively low strains. F-E models using composite interply fracture energy parameters, for example, G_{Ic} and G_{IIC}, are used to estimate onset of catastrophic damage growth. At this stage, these are suited only to the study of simple delaminations and cannot deal with the complexity of a real impact zone that includes broken fibers and matrix cracks as well as delaminations.

As yet, therefore, no model is sufficiently well developed for use as a predictive tool, for example, as linear elastic fracture mechanics (LEFM) is used as a predictive tool to estimate the residual strength of metals with cracks. Efforts

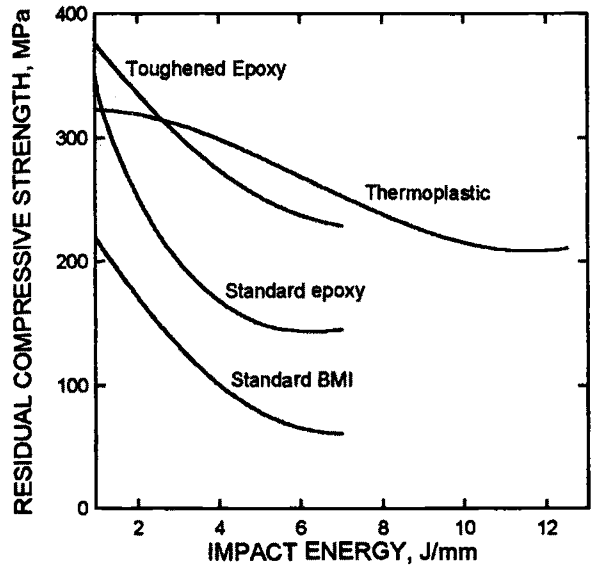


Fig 8.19 Residual compression strength versus impact energy for carbon-fiber composites with four common matrix systems. Taken from Ref. 2.

have increased over the past 10 years to develop predictive capabilities for both the consequences of the impact event (degree and characterization of damage) and residual static strength.

Software such as PAMSHOCK and LSDYNA is able to simulate the elastic response to the impact event; however, to characterize damage, multi-mode failure criteria have to be established for the material, lay-up, and configuration. This requires a significant amount of material testing and does not yet lead to an economic means of certification for typical structures.

Generally, the strength reduction in composites is determined empirically as a function of damage area and type. Most designs are based on the presence of an assumed damage zone for carbon composites; these issues are discussed further in Chapter 12.

Finally, cyclic loading and hot/wet environments have an influence on residual strength with BVID. These issues are discussed in the following sections.

8.8 Fatigue of Composite Laminates

In addition to maintaining static strength in service, structural composites are required to maintain an acceptable level of strength under fluctuating stress conditions, as experienced in service. The ability to maintain strength under cyclic stresses is called fatigue resistance. In an aircraft wing and empennage, the cyclic stresses are generally highly variable within the design limits; however, in fuselages, where the main stresses result from internal pressurization, the stress cycles to approximately constant peak values. These two types of loading are, respectively, called spectrum and constant amplitude.

In testing for fatigue resistance, there are two basic forms of measurement. The first is simply the life-to-failure (or to a certain level of stiffness degradation) at various stress levels; this is the S-N curve, where S is stress and N number of cycles. The second form is the rate-of-growth of damage as a function of cycles at various levels of stress. For metals, the damage is a crack; for composites, it is delamination or a damage zone consisting of localized microcracking and fractured fibers.

The ratio between the minimum and maximum stresses in constant amplitude cycling is an important parameter called the R ratio and is given by $R =$ minimum/maximum stresses. Thus, an R of -1 is a cycle that involves full reversed loading, $R = 0.1$ is tension-tension, and a large positive value, for example $R = 10$ compression/compression. The ratio R generally has a marked influence on fatigue resistance.

8.8.1 Tension-Tension Fatigue, $R \sim 0.1$

The tension-tension fatigue properties of unidirectional composites having high fiber/volume fractions are dominated by the fatigue properties of the fibers. However, fiber-to-matrix-stiffness ratio is also important, as the matrix is fatigue sensitive. If the fiber-to-matrix-stiffness ratio is not sufficiently high, the strain in the matrix can become critical. Provided the matrix is cycled below its strain limit for a given number of cycles, it will not be expected to experience fatigue cracking. Above this strain level, microcracking of the matrix will occur. Note that, due to constraint by the fibers, this strain level may be higher than the fatigue strain limit of the bulk matrix. However, the residual stresses resulting from the mismatch thermal expansions and Poisson ratio between the fiber and matrix are superimposed on the external stresses, which complicates the stress state in the matrix.

When a fatigue-resistant fiber such as carbon is loaded to a high percentage of its average ultimate stress or strain, some fibers with relatively large flaws or defects will fail, and the adjacent fibers will be more highly stressed over the region of the load transfer length. If the composite is unloaded and reloaded, a few more fibers will fail in these regions. Thus, when this is repeated over many thousands of cycles, a definite fatigue effect is observed, as shown in Figure 8.20 The S-N curve is relatively flat, but scatter is very high; this is a major feature of carbon-fiber-based polymeric matrix composites, particularly when subject to tension-tension fatigue.

If, however, the fiber itself exhibits a degradation of strength under cyclic loading, then a much more pronounced fatigue effect is observed. For example, glass fibers show a pronounced degradation under cyclic straining at high proportions of their ultimate strain, which is probably more related to cumulative

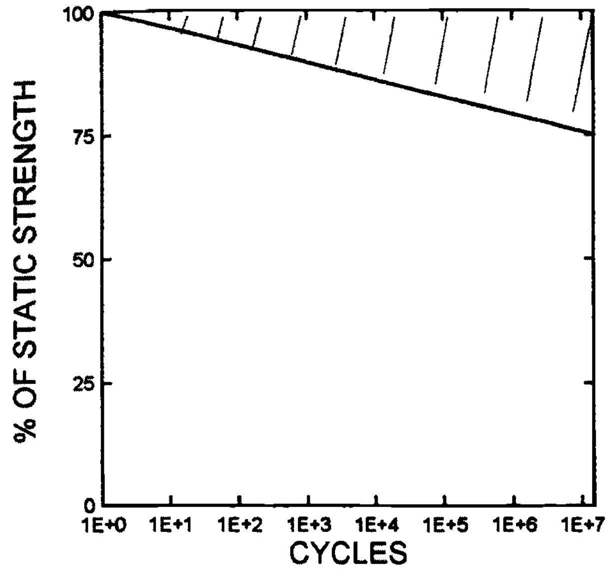


Fig 8.20 Typical scatter band for a unidirectional carbon-fiber/epoxy composite subjected to tension-tension cycling. Taken from Ref. 2.

time at high strain levels (stress rupture) than to damage caused by the cyclic loading. In addition, because of the low modulus of glass fibers, the resulting higher matrix strain results in matrix cracking that exacerbates fatigue sensitivity in two ways, first by strain concentration and second by allowing access of the environment to the fiber surface. As mentioned earlier, glass fibers are degraded by contact with moisture. To minimize stress on adjacent fibers, it is highly desirable that the fiber disbond from the matrix when fracture occurs. Similarly, when fiber fractures accumulate in a region, it is desirable that this region become isolated from the bulk of the composite by the formation of more macro-scale delaminations. Thus, composites with well-bonded tough matrices often exhibit inferior fatigue properties to those with brittle matrices.

8.8.1.1 Fatigue-Life Diagrams. Talreja[10] developed fatigue-life diagrams to explain the behavior of unidirectional composites under tension-tension cycling. Figure 8.22 shows a schematic diagram of a typical fatigue-life diagram for a carbon-fiber/epoxy composite that is divided into three regimes corresponding to different types of fatigue damage:

(1) Region 1 occurs at high stress levels and is a scatter band for failure of the fibers and therefore is centered on the strain-to-failure of the fibers. In this region, random fiber breaks occur at flaws on each loading cycle and may subsequently focus stresses on surrounding fibers. If disbonding of the

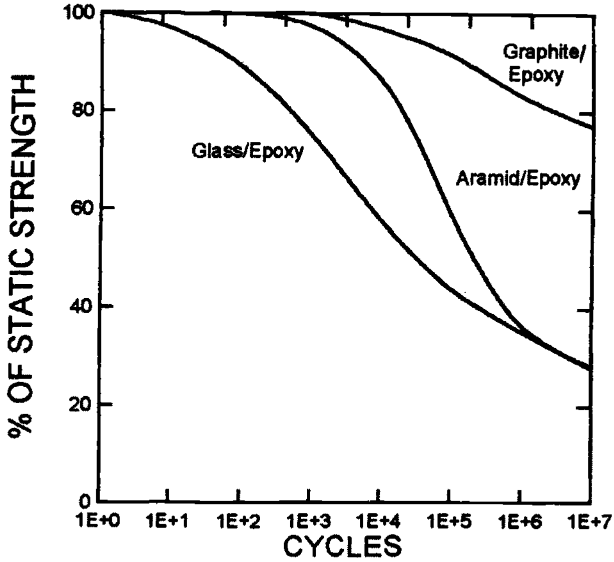


Fig 8.21 Schematic plot of S-N curves for unidirectional composites based on carbon, aramid, or glass fibers subjected to tension-tension cycling. Taken from Ref. 2.

broken fibers does not occur (because fiber/matrix bond strength is high), matrix cracks will form, increasing stresses on surrounding fibers. Even if debonding does occur, the accumulation of breaks in any cross-section increases net stresses, increasing the rate of random fiber fractures.

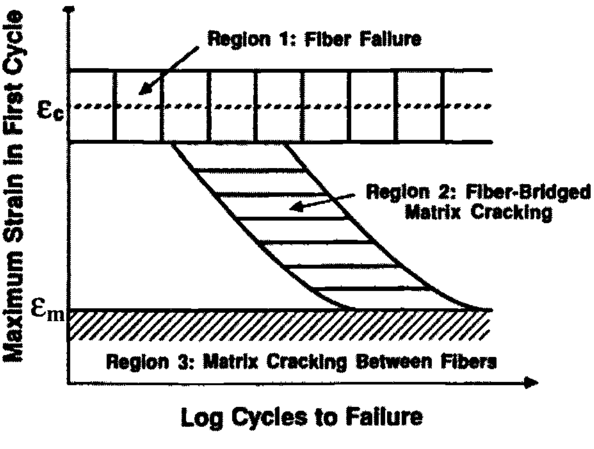


Fig 8.22 Schematic representation of a fatigue-life diagram showing damage zones for a unidirectional carbon-fiber/epoxy composite tested under unidirectional loading. Taken from Ref. 2.

(2) Region 2 is a region where cumulative matrix cracking and fiber matrix debonding occur. If debonding does not occur, the matrix cracking may result in fiber fractures, particularly if they impinge on a fiber flaw. Otherwise, the fibers are left bridging matrix cracks and will eventually fracture, as a bundle of fibers is weaker than those bonded as a composite (as explained in Chapter 2).
(3) Region 3 is below the fatigue limit for the composite because the strains are less than ε_m the nominal fatigue strain limit for the matrix. In this region, some matrix cracking may occur because of the local high thermal stresses and stress concentrations, but the cracks are non-propagating and therefore do not damage the fibers. However, cracking in this region will allow environmental ingress and lead to degradation in systems with environmentally sensitive fibers, such as glass fibers.

8.8.2 Tension Fatigue of Cross-Ply Composites

As may be expected, the fatigue behavior of cross-plied laminates is more complex than is the case for laminates with unidirectional fibers. This is because the off-angle plies are cyclically strained at some angle to the fiber direction, at a strain level that is largely dictated by the 0° fibers. As with static strength, microcracking of these plies between the fibers (FPF) can result in local strain elevation in the critical 0° fibers. Even if cracking of these plies does not result in failure of the 0° fibers, it is undesirable because it reduces the integrity and stiffness of the composite, even though the loss in stiffness is often fairly minor.

As cycling proceeds, the cracking pattern continues to accumulate until it saturates at what is called the characteristic damage state of the composite. Perhaps the greatest concern with this damage is that it opens the composite to ingress by the environment.

In laminates with fatigue-insensitive carbon fibers; provided there is a sufficient proportion of 0° fibers in the laminate, fatigue behavior is similar to unidirectional material—in other words, a flat S-N curve and high scatter. If the damage parameter is based on cyclic strain, the S-N curves could be fairly similar. However, with a low fraction of 0° fibers, in a quasi isotropic laminate, for example, fatigue sensitivity will be more marked with failure of the cross-plies, both reducing stiffness and concentrating strain on the 0° plies.

Three phases can be identified in the fatigue process, as illustrated schematically in Figure 8.23:

(1) Matrix cracking in 90° plies and, to a lesser extent, in the other non-zero plies is the first phase, which may initiate from the first load cycle depending on stress level—this is the FPF and will reduce laminate stiffness as the cracking accumulates. Eventually a characteristic damage state (CDS) develops as the cracking saturates.

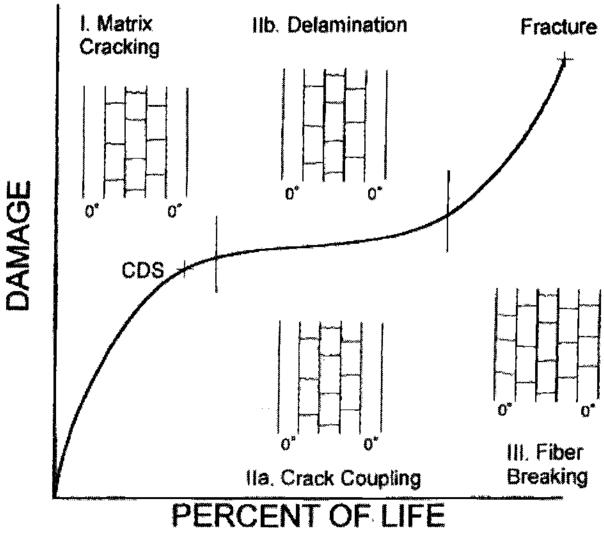


Fig 8.23 Schematic plot of fatigue damage mechanisms in a composite laminate. Taken from Ref. 2.

(2) Further cycling will result in the interaction and coupling of the matrix cracks through disbonding and the formation of delaminations. Depending on the construction of the laminate, edge delamination can also occur because of the high interlaminar stresses. The $[0°/\pm 45°/90°]_s$ is an example of an edge-delamination prone laminate. In this case, the edge plies become decoupled, and the delaminations propagate into the laminate, resulting in a marked elevation of the stresses in the 0° plies.

(3) As the off-angle plies become ineffective in carrying load, the stresses in the 0° plies will gradually increase and may cause accumulation of fiber fracture and eventual failure. The stress level for damage leading to final failure depends strongly on the volume fraction and stiffness of the 0° fibers, because, at a given stress level, they control the strain experienced by the composite and the load that can be carried when the off-angle fibers become ineffective.

Figure 8.24 schematically shows the reductions in residual strength and stiffness in a cross-ply laminate resulting from damage accumulation during cyclic loading.

8.8.3 Effect of Stress Concentrations

Stress raisers such as fastener holes and cut-outs are a feature of many composite components. Although these features often have marked detrimental effects on static strength in the as-manufactured component, they may not be a concern under cyclic loading. This is because the formation of minor

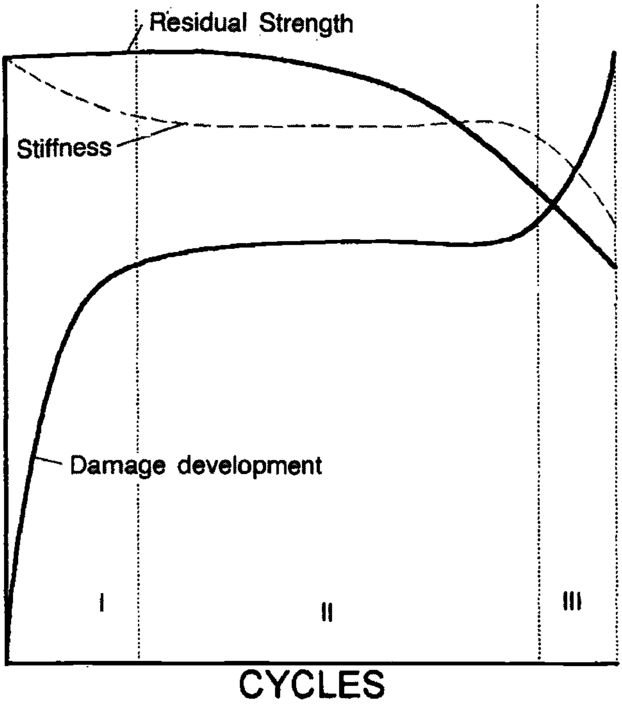


Fig 8.24 Schematic illustration of the changes in stiffness and residual strength as fatigue damage accumulates. Based on Ref. 2.

microcracking and delaminations in the high-strain regions can markedly reduce the stress concentration. This issue is further discussed in Chapter 12. It is, however, of interest to note the marked contrast in behavior with metals, where local plastic flow eliminates stress concentrations under static loading but results in fatigue cracking when the loading is cyclic.

Stress raisers also arise at ply-drop offs and at the ends of bonded or integral stiffeners. These features cause elevated through-thickness stresses transverse to the plane of the reinforcement. In this region, the rather poor fatigue resistance of the matrix, the ply/ply interface, and the fiber/matrix interface controls fatigue life. This situation is obviously highly undesirable, therefore considerable effort is made to minimize interlaminar stresses by careful design and test evaluation.

8.8.4 Effect of Loading Frequency

Unlike glass-fiber composites, discussed earlier, carbon-fiber composites with similar fiber architecture and matrix material are much less prone to temperature rise when loaded at high frequencies (5–10 Hz). This is because the high stiffness of the fibers limits the cyclic strain experienced by the matrix at a given stress level and, in addition, the high thermal conductivity of the carbon fibers conducts heat away from hot spots. However, for matrix-dominated composites (e.g., with $\pm 45^\circ$ fibers) matrix hysteresis is much greater, and significant temperature increases can occur. Other factors that have a major influence on matrix temperature rise include the properties of the matrix material and the thickness of the composite.[2]

8.8.5 Compression Fatigue, $R \sim 10$

The main concern regarding compression loading is the strength degradation caused damage such as BVID, which is discussed in the next section. In the absence of damage and providing the laminate has a high proportion of fatigue-resistant 0° plies, compression fatigue can be quite good under ambient conditions and can be similar to that for $R \sim 0.1$. Fatigue behavior is dependent on the ability of the matrix and fiber/matrix interface to suppress microbuckling of the fibers, which in turn depends on the resistance to microcracking under the negative loading. Because microcracking should be suppressed under compression cycling, quite good fatigue properties can be expected. However, compression fatigue sensitivity under hot/wet conditions will be more marked than under tension because of reduced support of the fibers by the softer matrix. Markedly reduced compression fatigue resistance is to be expected at temperatures near to T_g when the matrix softens.

Finally, as may be expected and because of their low resistance to compression loading, aramid-fiber composites exhibit quite poor compression fatigue properties.

It is of note that it is quite difficult to conduct fatigue tests under high compressive stresses because of the need to suppress global buckling of the composite. To achieve this, the use of anti-buckling guides is generally required, which allows only a small proportion of the cross-section to be tested. Thus, tension and compression results quoted may not always be directly comparable.

8.8.6 Tension/Compression Fatigue, $R \sim -1$

The influence of a negative R ratio on fatigue resistance of a unidirectional carbon/epoxy composite is adverse. The reason for the poorer performance under reversed cycling is that the damage caused to the matrix and various interfaces during the tensile cycle limits the ability of the laminate to support the fibers against buckling in the compression cycles. This behavior is evident in both unidirectional and cross-plied laminates.

8.8.7 Effect of BVID on Fatigue Strength

Unlike glass- and aramid-fiber composites, in which fatigue strength for undamaged structures may be a concern, fatigue of carbon-fiber composites is only a real concern when the laminate also contains low-level impact damage (BVID). Under these circumstances, there is a gradual reduction in residual strength with cycles.

The effect of BVID and fatigue on the compressive residual strength of carbon/epoxy composites is shown in Figure 8.25, taken from Ref. 11, which schematically shows the reduction in normalized residual strength that occurs as a function of impact damage size, and the further reduction caused by strain cycling at various normalized strain levels. The static failure strain plateau of around 4500 microstrain shown, which occurs at a damage size of about 25 mm, is typical for carbon/epoxy composites after BVID. The Figure also shows that a further reduction of the plateau to around 3000 microstrain results after cycling at a strain of around 0.6 $\times$ 4500 microstrain for around 10^5 cycles.

Some further experimental results showing damage growth[12] are presented in Figure 8.26. These are for a 56-ply-thick (approximately quasi isotropic) laminate with BVID subjected to a compression-dominated spectrum loading typical of that for a fighter upper wing skin. Results for both ambient temperature dry and hot/wet are presented. The influence of environment is discussed again in Section 8.9 of this chapter.

The reduction in residual strength, which results from cycling under compression-dominated loading, is associated at least in part with growth of the BVID delaminations. Some observations on growth of the delamination damage from BVID in carbon-fiber/epoxy composites during cycling under representative spectrum loading[13] are shown in Figure 8.27.

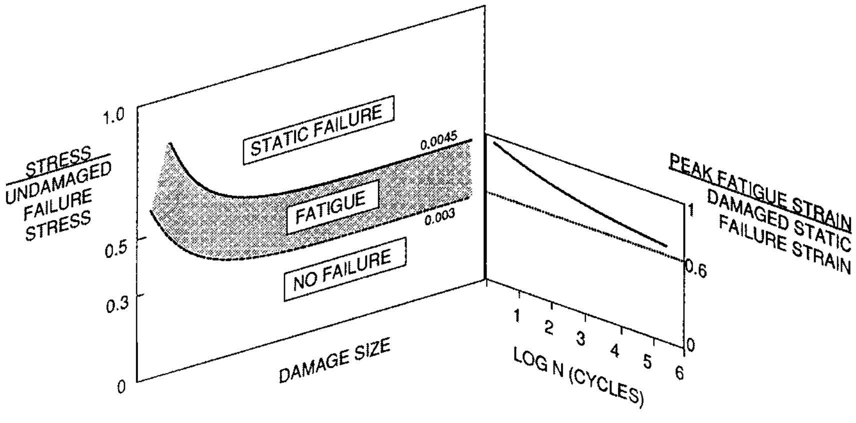


Fig 8.25 Schematic representation of the damage size on the fatigue life of composite laminates. Taken from Ref. 11.

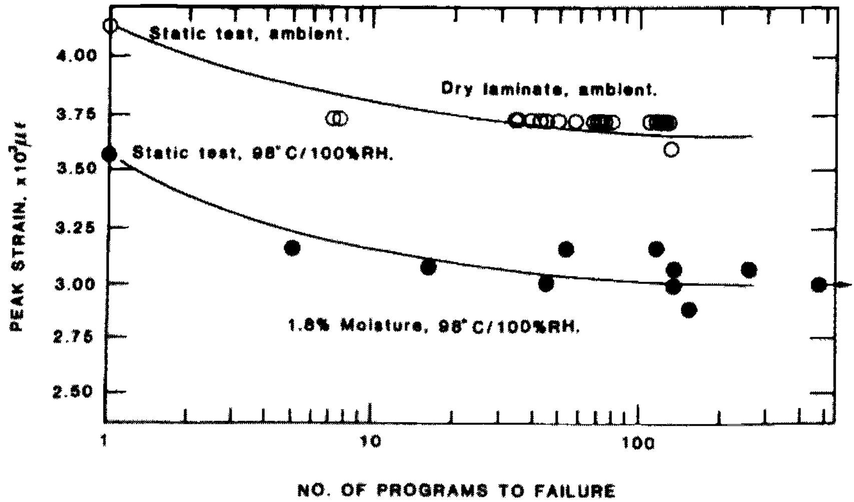


Fig 8.26 Compression residual strength for a 56-ply XAS/914C laminate as a function of spectrum loading representative of that experienced in a fighter wing skin. Behavior under ambient dry and hot/wet is presented. Taken from Ref. 9.

8.8.8 Damage (or Defect) Growth

Metal structures have long been designed with an understanding of the mechanics of flaw growth, allowing establishment of inspection intervals based on knowledge of the critical (unstable) flaw size and the growth under (spectrum) loading. The problem is much less well understood in composite materials. The common experience is that defects or damaged areas do not grow at repeated strain levels, even as high as 80% of limit strain, generally around 2000–3000 microstrain.

Figure 8.28 shows a comparison of residual strength reduction under spectrum loading (representative loading for the component) for metal and composite with,

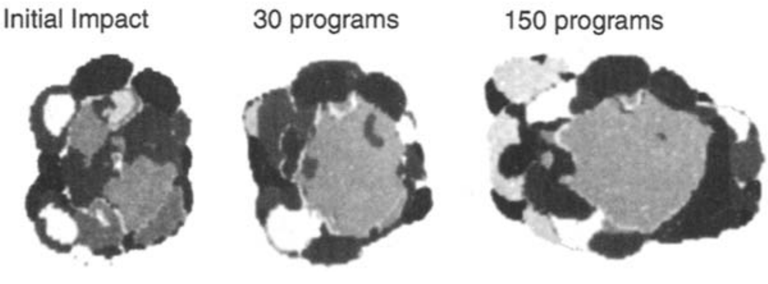


Fig 8.27 Growth of BVID delaminations under spectrum loading, representative of a compression-dominated wing top surface. Taken from Ref. 9.

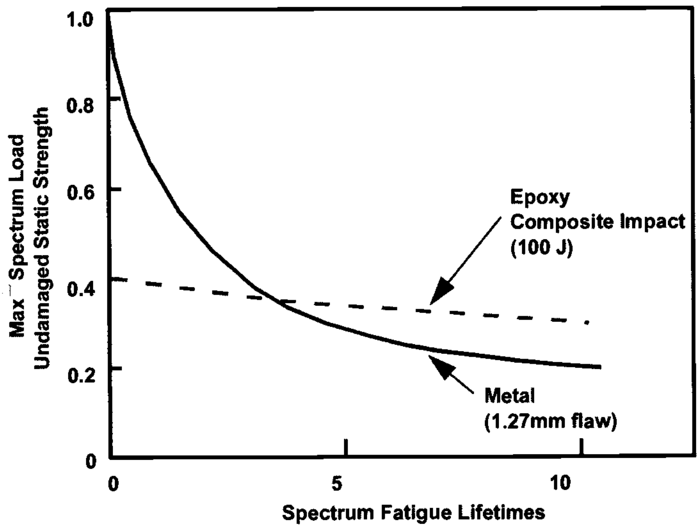


Fig 8.28 Residual strength as a function of spectrum cycles for 7075 T6 aluminum alloy and carbon/epoxy AS4/3501-6. Adapted from Ref. 14.

respectively, a crack-like flaw and impact damage.[14] In the metal, the flaw grows as a fatigue crack, causing a gradual reduction in residual strength until final failure. By contrast, the reduction in residual strength in the composite is marked immediately after impact. This reduction is followed by a relatively small further reduction in residual strength, with cyclic loading with or without limited flaw growth.

8.9 Environmental Effects

The mechanical properties of polymer-matrix composites may be degraded by a range of physical and chemical effects. It is therefore important to understand both the material and the operating environment. The environment must be understood before the suitability of a composite material for a particular application is assessed. The details include humidity, temperature (of air and material surface), ultraviolet and infra red radiation levels, wind conditions and, rainfall. Additional details that need to be defined specifically for flight include variations of these conditions with altitude, rates of heating and cooling during maneuvers, and frequency and duration of each exposure. The different mission profiles that the aircraft experiences will affect the severity and frequency of these conditions. Materials on different parts of the aircraft will experience different environmental conditions. A set of conditions or flight profiles must then be defined for each specific part of the aircraft. Aircraft components are also subjected to a wide range of chemicals, including fuel, fuel additives, hydraulic fluids, de-icing fluids, paint and paint strippers, dye penetrants, and ultrasonic couplant gels and detergents/wash systems.

These chemicals and environmental factors can significantly change the mechanical properties of a composite material. It is important to note that the matrix, fiber, and fiber/matrix interface will be affected differently by the operating environment. It is therefore vital to assess potential effects on each of these components with appropriate mechanical tests before certifying the material for use. Tests may need to be conducted to include the possible synergy between the various exposure environments. For example, the combination of humidity and temperature creates what is often termed a hot/wet environment. For epoxies and other thermosetting-based composites, the composite matrix absorbs moisture from humid air, which, when combined with elevated temperature, significantly reduces mechanical properties. Typically, for a carbon/epoxy composite cured at 180°C, moisture uptake reduces matrix-dominated mechanical properties (such as compressive strength) by up to 10% for subsonic aircraft and up to 25% for supersonic aircraft.

8.9.1 Moisture Absorption

Diffusion of moisture into a carbon/epoxy composite occurs quite slowly at ambient temperatures but will eventually saturate the material to an equilibrium concentration of moisture which will depend on operating conditions. It is important to understand the diffusion of moisture into the composite so as to allow the prediction of the long-term moisture content of a component in service. This moisture-content level can then be reproduced in the laboratory, and the effects on performance can be evaluated using mechanical tests.

The effect of diffused moisture on the mechanical properties of a composite is often reversible upon drying of the material. The absorption and desorption of moisture will occur on a continual basis for the life of a component in service as it is exposed to a changing environment. In the long term, the bulk of the material will come to moisture-content equilibrium, whereas the exposed outer surface plies may have a fluctuating moisture concentration.

Immediately after manufacture, the matrix material in a composite material will begin to absorb moisture from humid air, even in an air-conditioned environment. The effect of this absorbed moisture on mechanical performance is critical and needs to be evaluated for specific composite systems. The level and extent of moisture diffusion are highly dependent on the type of matrix material. As discussed in Chapter 4, thermoplastic resins absorb very low levels of moisture (typically much less than 1% by weight) while epoxy resins absorb over 4% moisture by weight. Therefore, it is not valid to assess all composite matrices to a fixed moisture content, but rather, the approach should be to expose the matrix to a representative environmental condition and assess the effects on the material over time. It is assumed in most cases that the fibers do not absorb moisture. This is true for carbon, boron, and glass but not for aramid and other polymeric fibers.

Moisture diffusion characteristics are often quoted in the literature as *Fickian* or *non-Fickian.* Fickian behavior, the simplest to deal with mathematically,[15] fortunately represents the diffusion behavior of most aerospace, thermoset matrix materials. It is characterized experimentally by an uptake of moisture that reaches an asymptotic value after a period of time (Figure 8.29). If the weight gain of a sample exposed to constant humidity and temperature is plotted against the square root of time, there will be an initial linear region up to about 60% of the maximum moisture uptake followed by a gradual approach toward a constant or asymptotic value. This asymptotic value is equal to the concentration of moisture experienced at the surface layer of the material exposed to a given humidity level.

Considering a common aerospace composite in laminate form, diffusion will occur primarily through the laminate faces, and only a small effect will be seen from edge diffusion. Fick's first law states that the flux of moisture in the through-thickness direction x, will be dependent only on the moisture concentration gradient through the sample in that direction:

$$MoistureFlux = -D \cdot \frac{\partial c}{\partial x} \tag{8.1}$$

where D is the diffusivity or diffusion constant, and c is the concentration of moisture. Fick's second law defines the differential equation for the diffusion process if diffusivity, D, is independent of x:

$$\frac{\partial c}{\partial t} = D \cdot \frac{\partial^2 c}{\partial x^2} \tag{8.2}$$

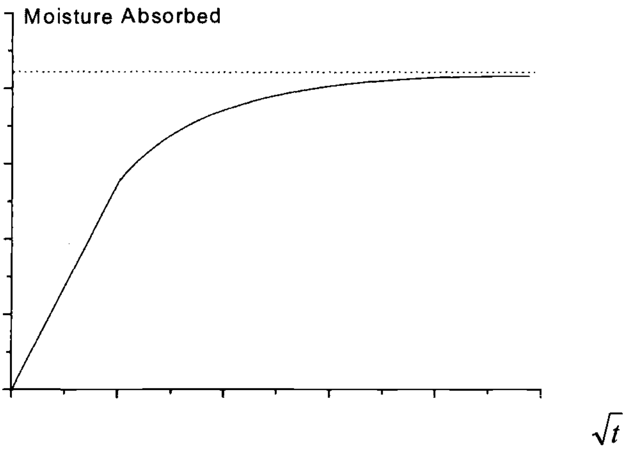


Fig 8.29 Schematic illustration of moisture uptake versus root time under constant humidity and temperature conditions for Fickian diffusion behavior.

The diffusion constant D will be independent of time and assumed to be constant through the thickness of the sample. The diffusion constant, however, varies exponentially with temperature:

$$D = D_o \cdot \exp[-E_d/R.T] \qquad (8.3)$$

where T is temperature, R is the universal gas constant, and D_0 and E_d are constants.

The absorption of moisture through the thickness of the composite with time is shown in Figure 8.30. After manufacture, the composite is essentially completely dry (t_0). Exposure to humid air allows moisture to begin to diffuse through the outer plies of the composite and through to the specimen center after time (t_1). After a longer period of time under constant humidity conditions, an even moisture distribution arises (t_∞)

It is important to note that at stages other than the fully dry or fully saturated case, a profile of moisture concentration will exist in the matrix. Because moisture affects the mechanical properties of the matrix, this also implies that a profile of properties will exist through the thickness. If the humidity conditions are transient (as found in normal weather patterns), a complex profile of moisture (and hence properties) through the specimen thickness may result.

The moisture content of a composite material is typically quoted as a percentage by weight:

$$\text{Moisture uptake concentration (\%)} = \frac{\text{Weight (final)} - \text{Weight (initial)}}{\text{Weight (initial)}} \times 100$$

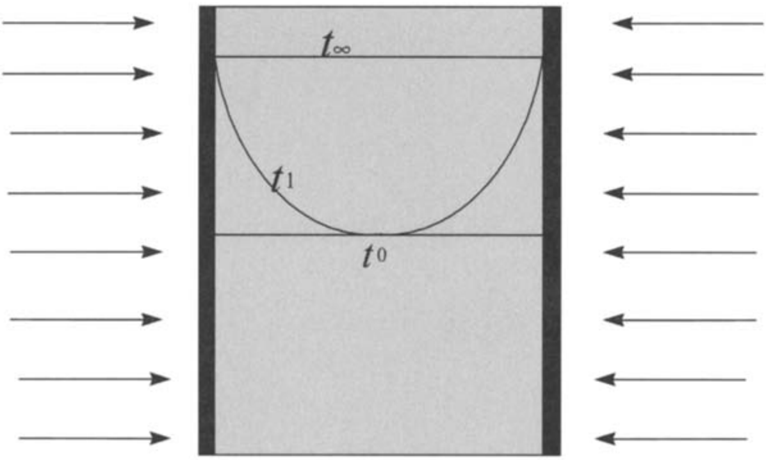


Fig 8.30 Diffusion of moisture into composite over time.

It has been shown for many composite materials that the maximum moisture uptake is related to humidity by:

$$\text{Maximum moisture uptake} = k.\phi^n \qquad (8.4)$$

where ϕ is the relative humidity in %, k is a constant, and n is close to unity for many aerospace composite materials. For a typical carbon/epoxy composite (60% fiber volume fraction), the maximum moisture uptake for exposure at 100% RH is about 2%. Figure 8.31 illustrates the influence of humidity on moisture absorption.

The diffusion of moisture into a composite matrix over time varies exponentially with temperature. An increase in temperature of 10°C typically doubles the diffusion rate. Figure 8.32 shows the effect of increasing temperature on the profile of weight-gain versus time for a constant humidity level.

The effect of specimen thickness is straightforward; the time taken to reach equilibrium is proportional to the square of the specimen thickness.

8.9.2 Real-Time Outdoor Exposure

The true performance of a composite material can only be established when all the exposures that will be encountered in service are allowed for. Real exposures to test coupons can be obtained by attaching the coupons to aircraft structure and assessing performance after differing lengths of service. This is somewhat inconvenient, and the alternative of simply exposing coupons of material to the environment at ground level is often substituted. Ground exposure coupons are more likely to absorb greater levels of moisture because they are unlikely to dry during flights at altitudes where the humidity is low. Consequently, this approach should be conservative.

The degradation of composite materials exposed on the ground to tropical conditions is described in Ref. 16. This trial used carbon/epoxy material as used

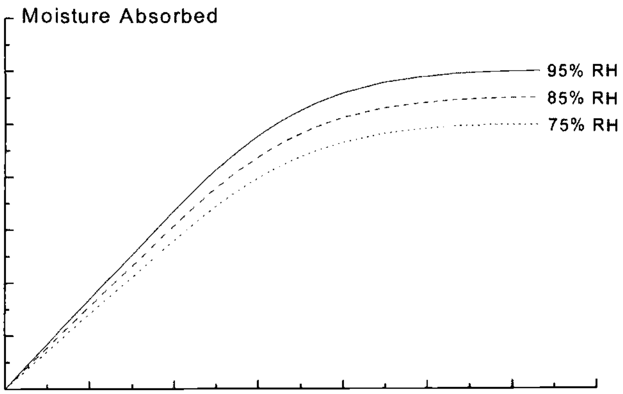


Fig 8.31 Effect of relative humidity on the amount of moisture absorbed with time.

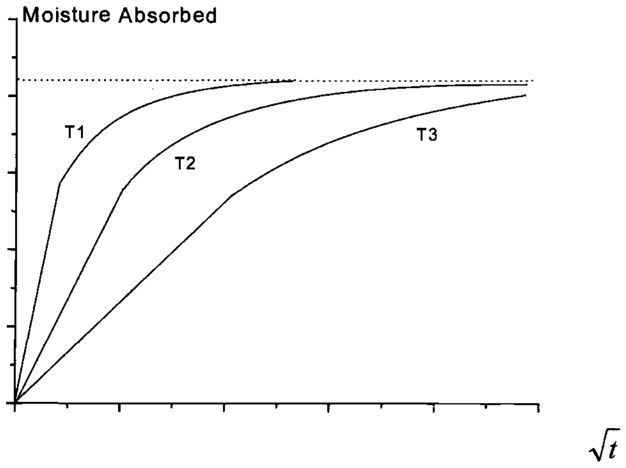


Fig 8.32 Effect of temperature on the weight-gain profile T1 > T2 > T3 (constant humidity).

on the F/A-18 aircraft. The trial investigated a number of variables, including different exposure conditions, the influence of paint schemes, and differing infrared and UV levels in an Australian tropical environment.

It can be seen (Figure 8.33) that it takes a very long time for the level of moisture to come to equilibrium, particularly if the sample is protected from sunlight. Note also that fully exposed (and to a lesser extent, shaded) samples lose mass due to erosion and UV degradation of surface matrix or paint layers. Typical equilibrium moisture contents for epoxy-based composites were found to range from about 0.7% to about 0.9% after long-term outdoor exposure.

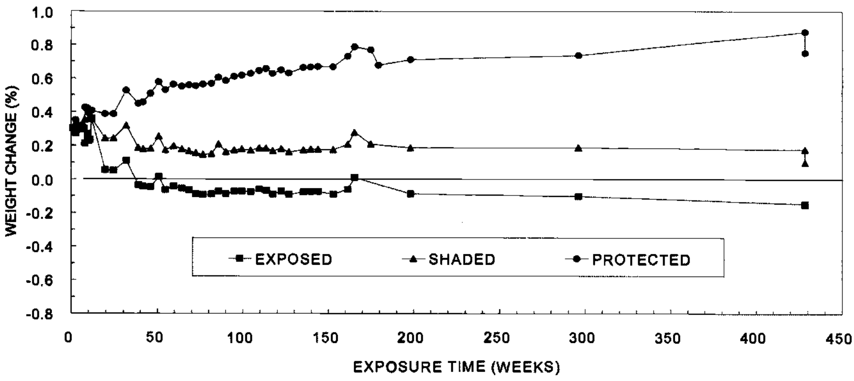


Fig 8.33 Weight changes for unpainted composite coupons (20-ply AS4/3501-6) exposed at Butterworth Air Force Base in Malaysia. Taken from Ref.16.

8.9.3 Effect of Moisture and Temperature on Mechanical Properties

Moisture plasticizes the matrix, leading to a reduction in T_g and changes in mechanical properties, such as Young's modulus. These properties can be restored on drying, assuming no permanent matrix damage, such as microcracks, occurred during the exposure. As discussed several times previously, the glass-transition temperature, T_g, is an important material property because it defines the temperature at which material properties are drastically reduced as the matrix changes from a glassy, stiff state to a pliable one. Figure 8.34 shows a plot of T_g versus exposure conditions for AS4/3501-6 carbon/epoxy composite.

As expected, the degradation in matrix properties mostly affects matrix-dominated composite properties at elevated temperatures. For example, Figure 8.35 shows that moisture has a marked effect on reducing the elevated temperature strength of $\pm 45°$ AS4/3501-6 laminates,[17] but little effect on the properties at ambient temperature.

The open-hole compression strength is often used as a key comparative measure of compression strength. Figure 8.36 plots the influence of temperature and temperature combined with moisture on this property for the AS4/3501-6 system and shows that moisture has a marked effect on elevated-temperature properties but, again, only a fairly small effect on the ambient-temperature properties.

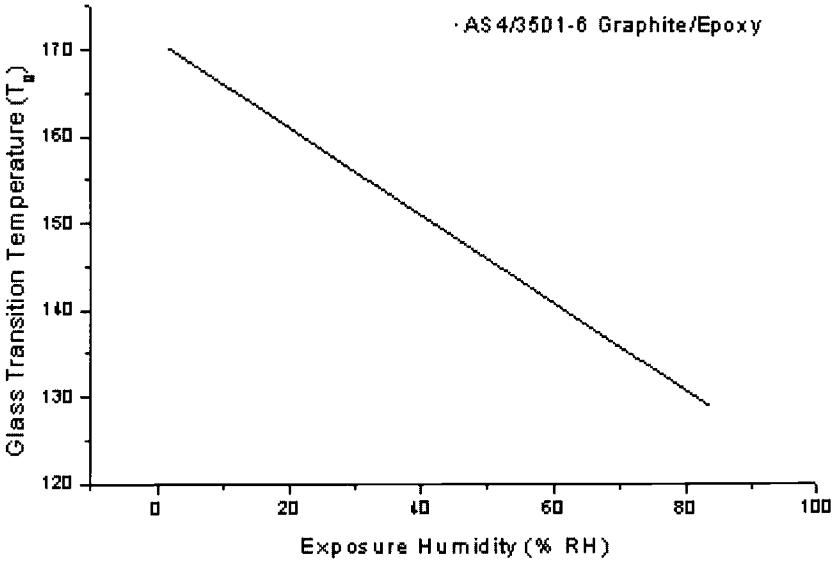


Fig 8.34 Plot of T_g versus saturation moisture content resulting from exposure to humidity for a typical carbon/epoxy composite.

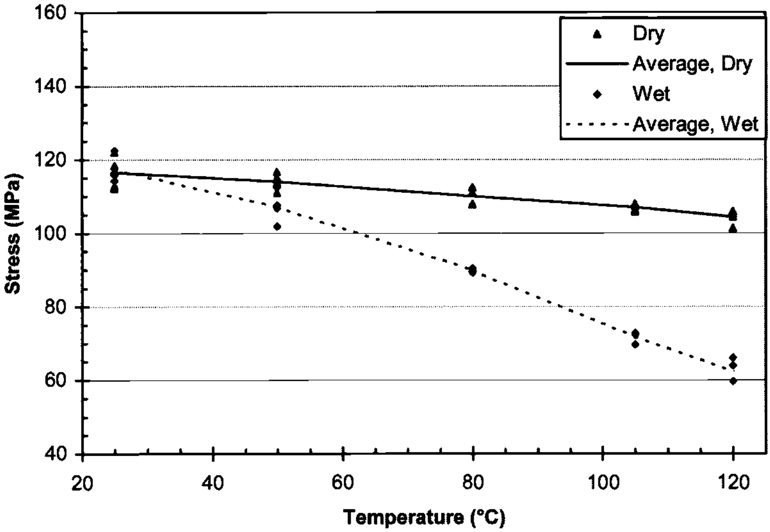


Fig 8.35 Effect of temperature and temperature and moisture of the tensile strength of $\pm 45^\circ$ laminates of AS4/3501-6. Taken from Ref. 17.

The environmental effects on fatigue behavior of composites for aircraft structures are as important as the effects on static properties.[18] Mechanical properties of the resin in carbon-fiber composites were not degraded after conditioning to a moisture level of around 0.9%. In fact, it was found that the

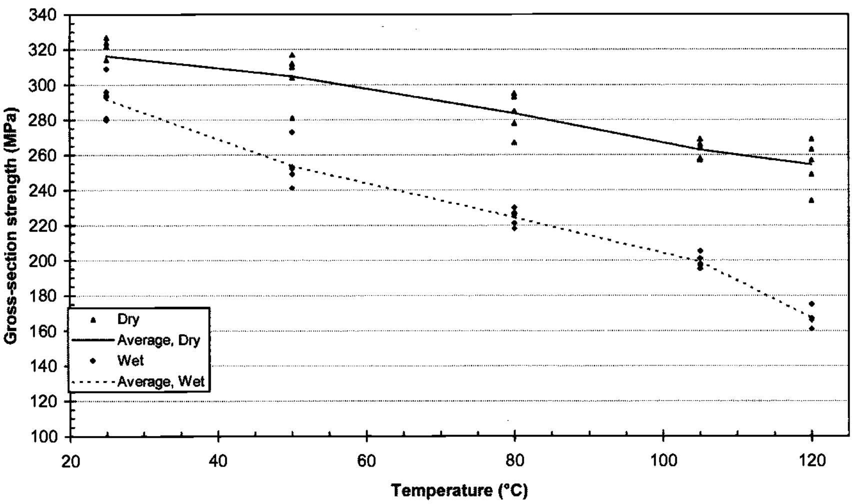


Fig 8.36 Effect of temperature and temperature and moisture open-hole compression strength of AS4/3501-6. Taken from Ref. 17.

interlaminar shear fatigue strength of the composite was moderately higher than dry, possibly due to the effect of plasticization of the matrix and a relaxation of residual thermal stress during processing. It was also found that $\pm 45°$ laminates conditioned to 0.9% moisture content had a better fatigue resistance than dry specimens.

The effect of absorbed moisture and temperature on the residual compression strength of impact-damaged specimens is shown in Figure 8.26 It seems likely, from the preceding discussion, that temperature rather than moisture causes the degradation, and the damage may be partially alleviated by the softening effect of the moisture.

8.9.4 Temperature Effects and Thermal Spiking

Both reversible and irreversible effects may be observed when a composite matrix is exposed to high temperature. If the temperature exceeds the T_g, the material modulus will decrease markedly. The original modulus will return upon cooling, provided no thermal decomposition or other permanent damage has occurred. Thermal decomposition is not an effect that is likely to be important on aircraft except on exposure to fire or other high-temperature sources (over 250°C). Irreversible damage effects include emission of volatiles and plasticizers, altering the nature of extent and cross-linking of the matrix and residual stresses induced through thermal cycling. Thermal cycling to high temperatures may cause trapped volatiles to expand and create high pressures within matrix voids and cracks. These stresses can lead to permanent mechanical damage and loss of properties. Service temperature values must be chosen such that the material wet T_g value is not exceeded (MIL-HDBK 17 recommends at least a 28°C safety factor).

Low temperatures can also have a serious effect on composite properties. Water trapped in voids or cracks within composite materials may freeze when exposed to cold conditions at high altitude. Temperature decreases by 3°C for every 500 m of altitude, and even moderately low altitudes can produce freezing conditions. Water expands by 8.3% in volume upon freezing and has a bulk modulus that is three to four times that of epoxy resin. Thus, water trapped within cavities will create pressure on the surrounding material that can result in permanent matrix damage. It is therefore important to consider the potential problems associated with free water entering voids and micro-cracks in composites. This can also be a more serious problem on a much larger scale for water entering honeycomb sandwich structures.

Thermal spiking is the terminology used to describe the effect of exposing a composite with high moisture content to rapid rises in temperature. Thermal spiking is an effect that is investigated to determine the useful maximum operating temperature of a composite. This is designed to simulate a supersonic dash in flight. Spiking is thought to create matrix damage that allows greater levels of moisture to absorb into the composite. Work by Clark et al.[19] shows that spiking creates the most amount of damage when conducted during moisture uptake or at the peak moisture uptake level. Thermal spiking before moisture conditioning did not appear to create any damage to the material.

8.9.5 Swelling Strains

The absorption of moisture by the composite matrix also causes it to swell and exert a resultant strain on the material, although, as mentioned above, these swelling strains may counteract residual strains from curing. The coefficient of moisture expansion defines the way a volume changes when taking up moisture. The volume change in a composite due to swelling may be represented by the empirical relation[20]:

$$\frac{\Delta V}{V_o} = 0.01 \cdot c + \frac{\Delta M}{M_o} \cdot d \tag{8.5}$$

where V_o and M_o represent initial volume and mass and c and d are swelling constants. For 3501–6 resin, swelling constants are given as $c = -0.61$ and $d = 0.87$.

8.9.6 Stress Effects

The effect of stress superimposed over a hot/wet condition can produce a greater detriment to composite mechanical properties than can each of these factors alone.[21] It is found that the level of moisture uptake increased with increasing stress level, and this behavior can be modelled using Fick's law. The greatest effect is on the value of diffusivity that increased with applied stress. The equilibrium moisture-content value is also found to be proportional to the applied stress level. Stress induced damage in the form of microcracks will increase the equilibrium moisture uptake level. Moisture present within these microcracks will have a significantly different effect than moisture absorbed by the matrix. Viscoelastic relaxations within the resin may occur at sustained stress levels that may alter the diffusion properties.

8.9.7 Ultraviolet Damage

Ultraviolet (UV) damage in composites occurs when the radiation breaks the chemical bonds in any exposed epoxy resin, and the degraded resin is then eroded away by wind and rain. Such damage can be minimized by the application of UV-resistant coatings. High-quality paints are equipped to deal with high UV levels and prevent the matrix from being affected. UV damage is restricted to the topmost layers of the composite and is unlikely to affect the bulk properties of a composite laminate. Other matrix resins, such as polyesters, are less susceptible to UV degradation, however, because they have inferior mechanical properties, these resin systems are not considered for structural applications on aircraft.

8.9.8 Chemical Changes in Matrix

Time may have an effect on the matrix because of chemical reactions that may occur on a long-term basis. The matrix may leach chemical components that will affect the T_g and mechanical properties such as fracture toughness. The reaction of moisture with some of the matrix chemical components may also need to be considered. These reactions are likely to be hastened by increased temperature. The long-term chemical stability of most aerospace thermoset resins is excellent under normal exposure conditions and therefore generally does not need to be determined, unless the suitability of the matrix for an unusually severe environment is a concern.

References

[1] Sims, G. D., and Broughton, W. R., "Glass Fibre Reinforced Plastics," *Comprehensive Composite Materials*, edited by A. Kelly and C. Zweben, Elsevier, 2000.

[2] Mall, S., "Laminated Polymer Matrix Composites," *Composites Engineering Handbook*, edited by P. K. Mallick, Marcel Decker, 1997.

[3] Baker, A. A., "Bonded Composite Repair of Metallic Aircraft Components: Overview of Australian Activities," AGARD Conference Proceedings 550, Conference on Composite Repair of Military Aircraft Structures, October 1994.

[4] Wardle, M. W., "Aramid Fibre Reinforced Plastics," *Comprehensive Composite Materials*, edited by A. Kelly and C. Zweben, Elsevier, 2000.

[5] Pigliacampi, J. J., "Organic Fibres," *Engineered Materials Handbook, Composites*, ASM, Vol. 1, 1987.

[6] Drzal, L. T., and Ishihara, S., "Aramid Fibre Surface Treatments to Improve Adhesion to Epoxy Matrices," Proceedings of 33rd SAMPE International Conference, November 2001.

[7] Smith, P. A., "Carbon Fibre Reinforced Plastics-Properties," Comprehensive Composite Materials, edited by A. Kelly and C. Zweben, Elsevier, 2000.

[8] Drzal, L. T., Herra-Franko, P. J., and Hoe, H., "Fibre-Matrix Interface Tests," *Comprehensive Composite Materials*, Vol. 2, edited by A. Kelly and C. Zweben, Elsevier, 2000.

[9] Clark G., and Saunders, D. S., "Morphology of Impact Damage Growth by Fatigue in Carbon Fibre Composite Laminates," *Materials Forum*, Vol. 15, 1991, pp. 333–342.

[10] Talreja, R., "Fatigue of Polymer Composites," *Comprehensive Composite Materials*, Vol. 2, edited by A. Kelly, and C. Zweben, Elsevier, 2000.

[11] Clark, G., and van Blaricum T. J., "Load Spectrum Modification Effects on Fatigue of Impact Damaged Carbon Fibre Composite Coupons," *Composites*, Vol. 18, 1987, pp. 243–251.

[12] Saunders, D. S., Clark, G., van Blaricum, T. J., and Preuss, T. E., "Graphite/Epoxy Laminate Coupon Testing Program," *Theoretical and Applied Fracture Mechanics*, Vol. 13, 1990, pp. 105–124.

[13] Saunders, D. D., and Clark, G., "Fatigue Damage in Composite Laminates," *Materials Forum*, Vol. 17, 1983, pp, 309–331.

[14]Whitehead, R. S., "Certification of Primary Composite Aircraft Structures," 14th Symposium of the International Conference on Aeronautical Fatigue (ICAF), June 1987.
[15]Crank, J., *Mathematics of Diffusion*, Oxford University Press, London, 1959.
[16]Chester, R. J., and Baker, A. A., "Environmental Durability of F/A-18 Carbon/Epoxy Composite," *Polymers and Polymer Composites*, Vol. 4, 1996, pp. 315–323.
[17]Callus, P., Vodika, R., Baker A. A., and Bannister, M., "The Effect of Temperature and Moisture on the Shear and Compression Behaviour of Three Carbon Epoxy Composites," Prceedings of ACUN 4, July 2002.
[18]Jones, C. J., Dickson, R. F., Adam, T., Reiter, H., and Harris, B., "The Environmental Fatigue Behaviour of Reinforced Plastics," *Proceedings of the Royal Society*, London, Vol. 396, 1984, pp. 315–338.
[19]Clark, G., Saunders, D. S., van Blaricum, T., and Richmond, M., "Moisture Absorption in Graphite/Epoxy Laminates," *Journal of Composites Science & Technology*, Vol. 39, 1990, pp. 355–375.
[20]Delasi, R., and Whiteside, J. B, "Effect of Moisture on Epoxy Resins and Composites," *Advanced Composite Materials: Environmental Effects*, ASTM STP 658, edited by J. R Vinson, 1978, pp. 2–20.
[21]Gillat, O., and Broutman, L. J, *Advanced Composite Materials: Environmental Effects*, ASTM STP 658, edited by J. R. Vinson, 1978, pp. 61–83.