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Gamstedt, Kristofer; Andersen, Svend Ib Smidt

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Fatigue Degradation and Failure of Rotating Composite Structures – Materials Characterisation and Underlying Mechanisms

E. Kristofer Gamstedt and Svend Ib Andersen
Materials Research Department

Abstract The present review concerns rotating composite structures, in which fatigue degradation is of key concern for in-service failure. Such applications are for instance rotor blades in wind turbines, helicopter rotor blades, flywheels for energy storage, marine and aeronautical propellers, and rolls for paper machines. The purpose is to identify areas where impending efforts should be made to make better use of composite materials in these applications. In order to obtain better design methodologies, which would allow more reliable and slender structures, improved test methods are necessary. Furthermore, the relation between structural, component and specimen test results should be better understood than what is presently the case. Improved predictive methods rely on a better understanding of the underlying damage mechanisms. With mechanism-based models, the component substructure or even the material microstructure could be optimised for best possible fatigue resistance. These issues are addressed in the present report, with special emphasis on test methods, and scaling from damage mechanisms to relevant material properties.

Contents

Preface

This review concerns fatigue damage development leading to final failure in composite structures in rotating applications. The aim is to establish a better link to the damage mechanisms of the composite material itself, in order to identify the crucial parameters that control fatigue sensitivity. In practice, empirical schemes are frequently used, where the macroscopic properties are mapped with respect to processing parameters and material composition. The present review focuses on methods to characterise fatigue degradation, both macroscopically for design and microscopically on the materials level, and on mechanistic investigations found in the open literature. With a mechanistic approach, considerable savings could be done both in terms of materials characterisation, and by more precise predictions in design. To this end, a better understanding of the relation between damage accumulation and failure in fatigue of composites is necessary.

As can be seen from this review, there are a number of reasons that are impending in a more reliable and optimised use of composites in rotating structures. Among them there is the relatively poor understanding in how the accumulation of damage influences the macroscopic degradation leading to failure. Furthermore, the lack of validated test methods for characterisation and prediction of properties needs to be resolved. With this input, useful models could then be formulated which would significantly improve the performance of these composite structures.

1 Introduction

1.1 Motivation

Fatigue failure is by far the most common type of failure in service loading of structural materials, including composites. It has been estimated that over 80% of service failures can be attributed to fatigue degradation (Dauskardt et al., 1993). Reviews on fatigue in composites have been compiled by e.g. Talreja (1993), Andersons (1994), Konur and Matthews (1989), and Stinchcomb and Reifsnider (1979). The present survey focuses specifically on rotating composite structures, in which fatigue is inherently an important issue. Applications are as diverse as rotor blades for wind turbines and helicopters, marine propellers, flywheels, paper machine rolls, etc. Despite the disparity of usage, the structures are almost exclusively made of fibre-reinforced plastics in the major load-carrying parts, and fatigue degradation and failure are the main failure modes that should be avoided by sensible design. Even though loading modes, constituent materials and lay-ups may differ considerably among these applications, a common approach to the fatigue problem can with advantage be advocated. A joint approach has the benefit of potential technology transfer. Particular focus has been placed on design methodologies for aerospace applications, where the end-use properties and reliability are most important. For large volume applications, as in e.g. the automotive industry, the cost issues are more important, and optimised materials processing have been developed. A mutual transfer of technology between different fields of applications speaks in favour of a more general approach, in this case to various composite structures that rotate, and are hence subjected to fatigue loading. Special attention is given to suitable test methods to characterise the fatigue behaviour under the relevant conditions. Furthermore, an understanding of the underlying mechanisms is necessary for a more precise improvement the composite structures. Means how to investigate and interpret fatigue mechanisms are considered.

In the aforementioned applications, composites based on either carbon fibres and glass fibres are mainly considered. Since both these materials show similarities in mechanisms, and can be approached with the same methods, they are not distinguished in particular unless necessary. For instance, despite the large difference in fibre stiffness, the influence of interfacial strength on fatigue properties in carbon and glass fibre reinforced plastics have been found to be identical (Gamstedt, 2000a). This can be explained by the finite set of operative damage mechanisms (fibre breakage, interfacial debonding, matrix cracking and yielding) which are present in different kinds of polymer matrix composites.

A scaling link is schematically illustrated in Figure 1. The purpose of this illustration is to show the link from the application, down through the length scales, to the microstructure, which in turn strongly depends on the processing conditions, and the choice of constituent materials. The fatigue degradation at each of the shown length scales depends on the active damage mechanisms in the antecedent smaller length scale. The performance of the structure thus depends on e.g. the buckling of the subcomponents, which in turn depends on the delaminations at the ply level. The fibre-matrix interface, and crack bridging bundles influence the delamination. In this way a qualitative link can be established from the underlying damage mechanisms all the way up to the performance of the structure. This requires careful experimental work on all length scales, and ample characterisation if the link should be made quantitative. With a quantitative link, we would have a mechanistic tool at our disposal with a potential to significantly improve the use of composites in these

types of structures. This philosophy has also been underscored in the approach suggested by Spearing et al. (1998), and Sørensen and Jacobsen (1998a).

Rotating composite structures in applications ~ 1 - 30 m

Wind turbines Flywheels Aeronautical rotors and propellers

Useful engineering design tool against fatigue of composite structures

Composite component subjected to fatigue loading ~ 1 m

Composite component

Prediction of structural response in fatigue loading

Growth of damaged zone eventually leading to ultimate fatigue failure ~ 10 cm

Damaged zone

Mechanistic modelling of damage growth

Underlying fatigue mechanisms responsible for propagation of the damage zone ~ 1 mm

Damage mechanisms

Microstructural tailoring to suppress deleterious fatigue mechanisms

Materials engineering: Role of composite constitutives and their interfaces in the active damage mechanisms ~ 10 μm

Micromechanisms

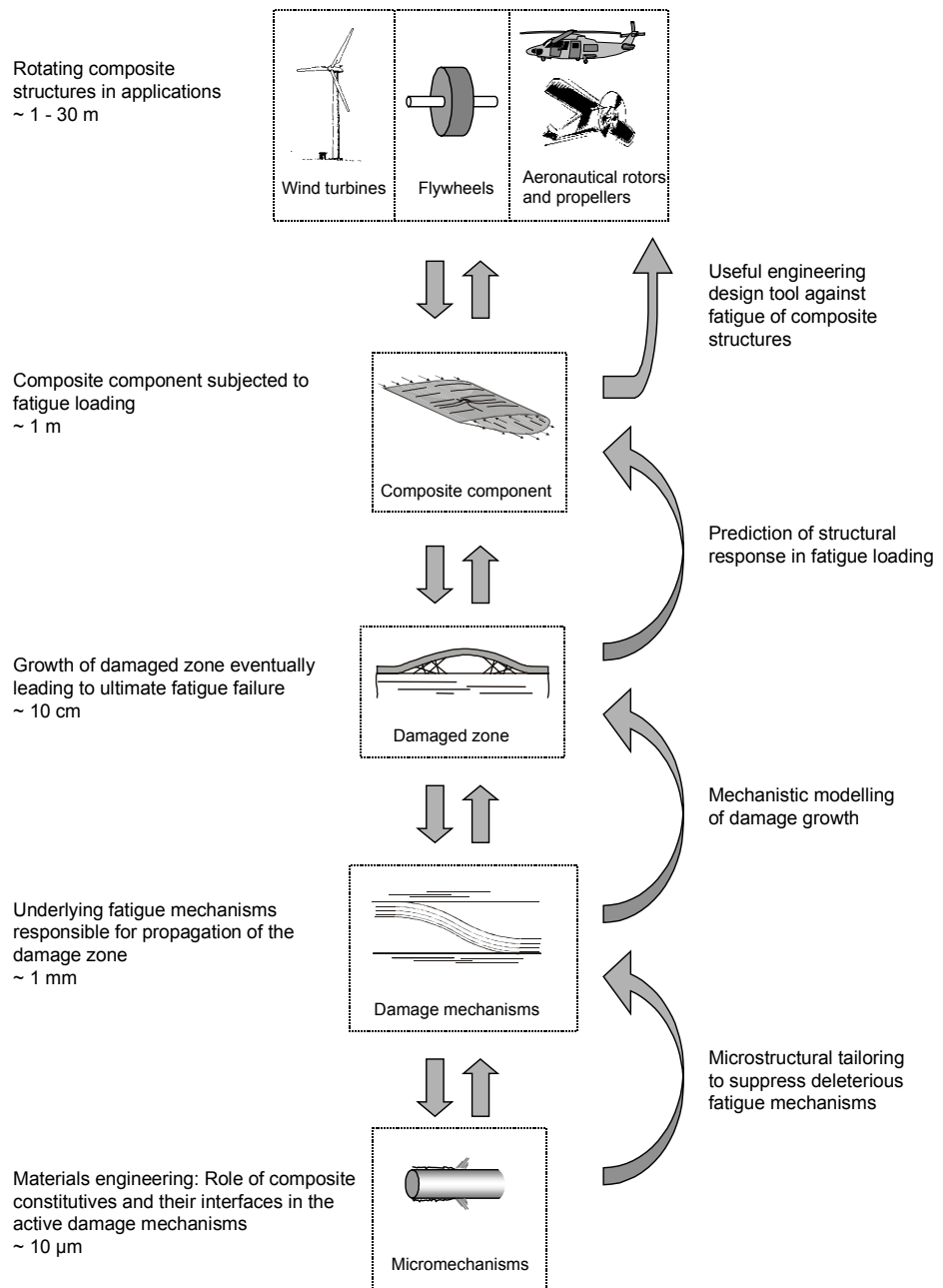*Figure 1. Schematic illustration of a qualitative link from applications with rotating composite structures through damage accumulation and failure induced by fatigue loading down to microstructural features of the composite.*

In the following, examples of applications with rotating composite structures are listed, and the specific fatigue considerations and commonality of these applications are highlighted.

1.2 Rotor blades

Composites rotor blades are with advantage used in various applications such as wind turbines, helicopters, air ventilation units, as well as aeroplane and marine propellers. Apart from high specific strength and stiffness, the fatigue resistance of composites is of utmost importance in these applications, since the loads fluctuate and the structures are generally designed to remain in service for many years.

Wind turbines nowadays are designed with rotor blades usually made of glass fibre reinforced plastics. These structures are safe-life constructions, i.e. their integrity should be guaranteed over the total foreseeable service life of the wind turbine, which is in general 20 years (Kensche, 1996). During this time the rotor blades are expected to endure more than 10^8 load cycles. In lack of comprehensive fatigue data for the material used, the blades have been designed at low and conservative design levels. Such allowables have led to an unnecessary high structural mass of the rotor, and thus also increased loads on the drive train, nacelle and tower. With lighter masses, larger and more effective turbines can be constructed. Lighter blades have also the advantage of reduced costs. Data on long-term fatigue properties must be proven. Mandell and Sutherland (1999) have compiled a database specifically on fatigue of composite materials for wind turbine blades. In wind turbines, the fluctuating loads arise primarily from gravity, since the typical size of the new rotor blades have increased considerably in the last few years. The gravitational edgewise loads are highest near the hub. With the anticipated larger structures the incorporation of carbon fibres to form hybrid structures together with the conventional glass-fibre-reinforced plastic will soon become reality, as a measure to maintain low flapwise deflections. Fatigue of these structures also arises from aerodynamical flapwise loads from wind speed fluctuations. The use of hybrid materials requires careful fatigue characterisation, since the fatigue behaviour can not be directly interpolated from data from the two parent materials (Peijs et al., 1990).

Fatigue failure in the composite or in adhesives or joints connecting the composite to other components in the helicopter rotor blades is a common type of failure (Espinosa and Groepler, 1992). Just as for other rotating structures, the fatigue resistance of helicopter rotor blades can be considerably improved by macrosopic structural optimisation to minimise vibrations (Yuan and Friedmann, 1998), which also includes sensible elastic tailoring (Smith, 1994). Special care has to be taken in the design of the tapered composite parts near the hub, including edge effects and structural corners (Chan, 1992). In parallel, by use of a fatigue resistant composite material, further safety against fatigue failure is achieved. Not only the rotor blades are susceptible to this type of failure, since rotor vibrations may also induce fatigue failures in adjacent composite parts of the fuselage (Overd, 1993).

Composite materials are gaining increased use as propellers in aeroplanes, which is partially due to their fatigue resistance (Bokulich, 2000). Other attractive features are the possibility of aeroelastic tailoring and the high specific strength and stiffness. As for helicopter rotor blades, aeroplane propellers should also be designed fail-safe, since failure can have devastating consequences in these applications. Since propellers are subjected to cyclic loading, design against fatigue failure is crucial also for this application.

Composite propellers are also be advantageously used for propulsion of marine vessels (Mer, 1994). The reaction forces exerted by the water are considerable, and are likely to be highly variable due to wave motion. Fatigue resistance is therefore of great concern. In addition to the obvious advantages of lightweight and absence of pronounced corrosion, composite propellers behave

very different from metals under impact. Damage is localised, and the integrity of the shaft and gearbox is not compromised, which is often not the case with metal propellers. Measurements have shown that the composite propellers are sufficiently stiff to perform at least as well as their metal equivalent in bollard pull, speed and open water efficiency. Design efforts are now generally directed towards evaluation of long-term properties of composites in a marine environment, i.e. fatigue loading, and towards the exploitation of the anisotropic properties of composites to increase efficiency.

1.3 Flywheels

Composites are gaining increased use in flywheel applications for energy storage. The advantage of the composites in flywheels lies in their high specific strength and high specific stiffness. Inertial loading causes stress in the material at high rotational speeds, and can be minimised by use of a suitable lightweight composite material. High strength is needed to achieve maximum rotational speed. Advanced composite rotors enable the storage of greater energy per unit weight or per unit volume compared with other materials. The power losses in the bearings can thus be reduced. Furthermore, fibre reinforced composite rotors have been shown to fail in a more benign manner than metallic rotors, which is an important factor for safety reasons. Flywheels offer a potential of higher power density compared with conventional energy storage devices. There is much interest in flywheels for energy storage in a number of diverse applications such as satellites, stationary daily storage, uninterruptable power supplies, and hybrid electric vehicles. Composites also offer some important advantages in the disc between the shaft and the rim, in particular in the presence of gyro forces such as in mobile brake/acceleration devices in vehicles (Andersen, 1998).

   The angular velocity of flywheels is generally so high that the gravitational forces can be neglected as source of fatigue degradation. At constant velocity, creep in the composite is the most important loading mode, but since the flywheels are regularly accelerated and decelerated in order to store and release the kinetic energy, stress variations are frequent. In a mobile environment, gyro effects contribute to additional variations in loading. A combination of creep and fatigue is therefore the most relevant loading mode to characterise composite materials for flywheel applications. In terms of testing, this means fatigue loading with a non-zero mean stress. The principal stresses are predominantly tensile, unless substantial compressive stresses have developed during fabrication. Flywheels can be designed to develop delamination by radial tensile stress in an overload situation, which is a preferred failure mode for safety reasons. The creep-fatigue properties in the weak material direction (perpendicular to the fibres) may thus be of importance in a design.

   At operational speeds of up to 75,000 rpm, a failure event could launch fragments of composite material at ballistic rates. In order to design a rotor for, say, 20 years of service, the designer must know how well the material will hold up after being subjected to many cyclic variations of speed and stress. Repeated charging and discharging cause stress variations, both in the hoop and radial directions, as well as in shear by torque. Temperature fluctuations induce fluctuations in thermal residual stresses experienced by flywheels without the benefit of temperature control (Gabrys and Bakis, 1997). Stationary flywheels are generally kept in vacuum to avoid air drag. Shirey and Bakis (2000) have investigated the fatigue properties of carbon and glass fibre composite materials intended for flywheel applications with and without vacuum conditioning and

tested in medium vacuum. Small or no significant changes in fatigue properties were found due to conditioning and vacuum testing.

1.4 Paper machine rolls

Various rolls in paper machines are nowadays being designed with carbon fibre reinforced plastic materials, which show the advantage of being relatively fatigue resistant, with high specific stiffness, and relatively resistant to corrosive environments. The increasing running speeds of paper machines in e.g. calendering are setting new demands on the fatigue resistance of the roll materials. Traditionally steel rolls have been used, which have been very costly to replace after fatigue failure. To design composite rolls, a deeper understanding of the fatigue behaviour at very high cycle fatigue ($> 10^9$ cycles) is necessary. Tension-compression is the main loading mode, and the presence of moisture and corrosive agents makes the analysis challenging (Lenz and Himmel, 2000). Furthermore, the viscoelastic nature of the polymer matrix composite has the advantage of a more gentle contact and a larger area of contact to the paper as compared with steel rolls (Vuoristo et al., 2000).

1.5 Outline

In the subsequent section, a review of useful test methods for structures is presented, and their shortcomings are highlighted. The plethora of different kinds of fatigue tests is condensed to a few common tests that characterise the materials in different types of loading, thereby invoking different failure modes and damage mechanisms seen in the structure. Some non-destructive test methods of complementary use to the mechanical tests are outlined, as well as some principles concerning post-mortem fractography. Subsequently, findings of the fatigue damage mechanisms in composites used in the applications in mind are presented in the following section. Emphasis is made on how the damage mechanisms influence the macroscopic failure, i.e. predominantly the degradation of stiffness and strength, which eventually leads to final failure. The next sections deals with predictive techniques for property degradation and lifetime assessment. Finally, the conclusions are drawn in the last section.

2 Fatigue testing

2.1 Fatigue testing of structures

Realistic fatigue testing of the entire composite structures or of relevant substructures is a necessary step in the validation of a newly designed application. How these tests should be carried out in detail is very dependent on the intended service use, which generally varies considerably from one product to another. The structural tests are usually carried out by the manufacturer, who has the necessary knowledge of the indented service conditions. The intricacy of large scale testing of structure or bulky components can be illustrated with a methodology used in testing entire rotor blades for wind turbines in fatigue. The blade is strapped up in a well thought-out construction by a whiffle tree, which are used mainly in static loading. The 'tree' structure distributes the load to a number of points, where all links and connections are compliant to prevent the

load fixture from effectively alter the blade stiffness. The geometry is designed to apply loads to the blade that approximate the distribution of the target test load, i.e. the forces generated by wind. A crane structure is commonly used to apply the load, although hydraulic actuators can be used for smaller structures. The point is that structural testing is highly specialised and aims to mimic the load conditions experienced in service, whereas specimen testing is performed under well controlled conditions to characterise the behaviour of the material. Rubin (1992) have reviewed common failure modes of composite aircraft loads, and suggested how to find material allowables by simplified tests. Typical failure types include (*i*) shear delamination in a skin, (*ii*) pull-through of fasteners in skin laminates, (*iii*) out-of-plane interlaminar tensile failure in corners, (*iv*) adhesive failure of bonded or co-cured spars, ribs or stiffeners, (*v*) flange crippling, and (*vi*) tensile failure in a web. All of these structural failure modes can be reduced to simpler tests of coupons or laboratory-scale elements, which would allow verification of structural capabilities. Appropriate reductions to design allowables can then be applied to the design of major components or full-scale structures.
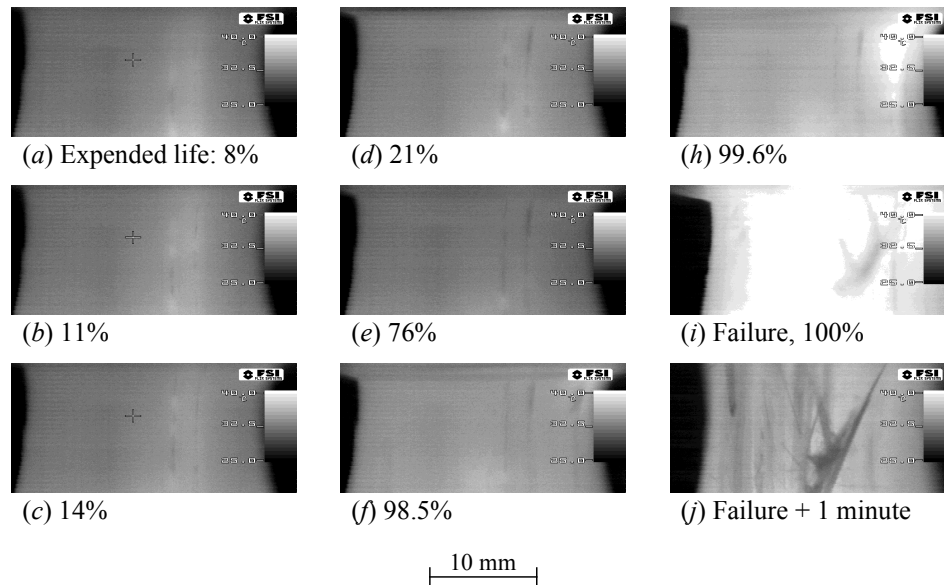
To test the entire structure for each small alteration in materials or design details is generally not feasible because of the high costs for structural tests. Instead, materials characterisation on the coupon level, or of small components in laboratory settings is common practice. The results from such tests can be used in design of larger structures if the relevant failure modes have been clarified and characterised. In this way, a large range of materials could be tested to rank their performance, and if done conscientiously, confidence can be lent also to design against fatigue with a proper life prediction method.

The scope of this review is not to cover important macroscopic tests of composite structures, but to give insight in the underlying reasons for fatigue failure. We will therefore have a closer look at some useful test methods to characterise the composite material on a coupon or specimen level in the following.

2.2 Tensile loading

Tensile testing fatigue testing of straight coupons is the simplest type of test in this context. The ASTM standard D3039 is frequently resorted to for tensile properties, and extended to cyclic tensile loading in standard D3479-76. The purpose is generally to generate *S-N* data for design or materials selection. As for all composites containing polymer matrices, it is important to maintain the frequency low enough to avoid extensive heating that would affect the test results. Experimental results show that a too high frequency gives shorter lifetimes in tensile fatigue, in particular for laminates containing off-axis plies such as [±45] (Kujawski and Ellyin, 1995; Lee and Jen, 2000). An off-axis orientation means increased shear stress in the polymer matrix. Cyclic shear of the matrix is the main dissipative mechanism in composites, and a large part of the mechanical loss is transferred into heat. Because of the low heat conductivity of polymers, and in particular for glass-fibre reinforced plastics on the composite scale, an adiabatic accumulation of heat may even result in temperature increase above the glass transition temperature of the polymer. In that case, one could expect a transition in mechanism and a radical increase in the degradation rate. Since damage gives rise to further stress concentration, meaning higher cyclic stresses, damage and heat will mutually increase in a self-escalating manner if the load frequency is too high. Jacobsen et al. (1998) have observed this phenomenon by means of thermographic imaging of a woven notched composite specimen subjected to tensile fatigue. In Figure 2, the

temperature evolution on the surface of a unidirectional is shown for an intermediate load amplitude with a total life of 64,300 cycles (Gamstedt et al., 2001c). The sequence of pictures shows that heat localisation means that impending failure awaits. In the white regions, the temperatures have risen in the excess of 20°C, getting closer to the glass transition temperature of the polyester matrix, which would mean a changeover to even more rapid degradation mechanisms. Naturally, it is of importance to check the specimen temperature every now and then during fatigue to assure that the applied ambient temperature is roughly the same as the intended specimen temperature. Otherwise, the interaction between heat and damage development would have to be taken into account in the characterisation of fatigue properties.

(*a*) Expended life: 8%	(*d*) 21%	(*h*) 99.6%
(*b*) 11%	(*e*) 76%	(*i*) Failure, 100%
(*c*) 14%	(*f*) 98.5%	(*j*) Failure + 1 minute

10 mm

Figure 2. Evolution of temperature field on the surface of a unidirectional 0° carbon-glass fibre hybrid composite (Gamstedt et al., 2001c). The load and fibre directions are vertical. Note that localisation in heat (and damage) sets in just imminently to failure, and then propagates rapidly across the specimen.

For multidirectional laminates, delaminations tend to grow from the edges and influence the subsequent accumulation of damage (O'Brien et al., 1989). Hence, the fatigue process becomes dependent on the specimen width, and does not reflect an intrinsic material behaviour. In a composite structure, the total edge area is small compared with the surface area of the laminate. This ratio is much larger for laboratory scale specimens. Rather than compiling stress-life data for multidirectional specimens that exhibit delamination growth during fatigue, one should characterise growth rates and the interaction with transverse cracking and other ply-level mechanisms. If these phenomena could be quantified and modelled, the stress history of the critical element of the laminate, typically the 0° ply, could be estimated. The expected life time for a more general geometry would then become predictable.

2.3 Compressive loading

In compressive loading, it is cumbersome to achieve a uniform stress field for composites, because of their high degree of anisotropy. If the load is applied in

the fibre direction, i.e. the principal stress direction is the same as the principal material axis, there will be a tendency for spurious failure mechanisms caused by specimen geometry and set-up of the test rig. The nature of gripping is known to influence the achieved compressive measurements (Joffe, 2000). The type of end tabs is also known to influence measured results, which obscures the true material properties (Adams and Finley, 1996). Of all possible test method, the one that gives the highest strength values or longest fatigue life provides the characterisation that lies closest to the material properties. A popular type of fixture is the modified Wyoming ITRII rig in which the load is transferred onto the specimen by shear along the grips (see e.g. Chaterjee et al., 1993). Still better would be a fixture that distributes the compressive end loads and shear loads along the end tabs by a hydraulic system in an appropriate ratio, which has been developed by Jacobsen (2001) based on the modified ITRII rig by Hsiao et al. (1995). A sensible application of load is particularly important for thick specimen, which tend to show lower strength values, not only because of a Weibull scaling effect in brittle failure, but mostly because of the dependence on the test itself as a result of gripping, macro-buckling in the absence of sufficient support, and effects of specimen geometry. Transverse compression is not so sensitive, because macro-buckling is less likely to occur in this case because of the perpendicular directions of the major material axis and the applied load.

The standard ASTM D695 is based on a specimen with a rectangular cross-section which is subjected to compressive loads at its ends. The method sometimes requires that supporting plates should be mounted along the sides to avoid macroscopic buckling. Instead ASTM D3410 and ITRII can be used, where the load is transferred by shear through glued end tabs. Another alternative is given by standard ASTM D5467 where the composite plates are glued on top and along the bottom of a sandwhich panel, which is subjected to four-point bending, whereupon one of the composite plates will be under compression, and the other under tensile loading.

The static methods may be generalised to fatigue simply by applying cyclic loads. However, relatively few studies seem to have been carried out on compressive fatigue, despite the relevance of this failure mode. Huang and Wang (1989) measured the *S-N* data for unidirectional polymer and metal matrix composites. These results were later correlated to a micromechanics model by Slaughter and Fleck (1993).

Partly because if the difficulties in obtaining material properties as stress-life, a more common approach is to investigate damage progression in a simplified component, e.g. a panel with a notch or a hole that resembles similar features in the intended structure. Soutis et al. (1991) have investigated the damage development in compressive fatigue of centre hole notched composite plates. Growth rates of splits and delamination emanating from the defects were characterised, which may be used to predict damage growth from other similar of defects.

2.4 Biaxial loading

Composite laminates in structures are very frequently subjected to biaxial stress, both on the ply level because of the anisotropy of the adjacent plies, and because of biaxial far field loading. Tri-axial stress states are less common for thin plates, since only in-plane loading is typical in designed structures. Biaxial fatigue testing can be achieved either by combined tension-torsion of filament-wound composite tubes (e.g. Inoue et al., 2000), or by loading cruciform specimens in two directions (e.g. Chen and Matthews, 1993). Just as for the

above load cases, measures should be taken to avoid geometry-dependent failure modes, such as initiatory cracks starting from the specimens boundaries. End effects are avoided in tubular specimens, but this specimen configuration requires special manufacturing, and may not yield the same kind of material, as the one used in the structure in mind. Cruciform specimens requires cautious manufacturing as well, since care should be taken to achieve the highest uniform biaxial stress in the central gauge region (Youssef et al., 1998). Before resorting to the cruciform test, one should convince oneself that the stress state is uniform in the gauge region, and that failure originates from within this region (Fawaz and Neale, 1995b). Special attention has to be paid to laminates with a high degree of anisotropy, and finite element analysis should be carried out to assure that failure can be anticipated within the gauge region in the centre of the specimen (Boehler et al., 1994).

2.5 Out-of-plane loading

Axial and uniform out-of-plane loading are in practice similar to transverse loading of a unidirectional composite. A far-field stress perpendicular to the fibre direction in out-of-plane loading is not very realistic since structures are designed to carry the largest part of the load in the plane. More importantly, laminate bending can give rise to a critical state. Even axial compressive loading is likely to result in out-of-plane deformation through bending and loss of structural integrity as e.g. delaminations develop and plies buckle. Specific delamination tests will be discussed in the following section.

Specimens with structural non-uniformities, such as ply drops, are likely to deform out of the original plane on loading, and mixed-mode crack propagation initiated from the ply drop is anticipated. Since fatigue delaminations often start from discontinuities like ply drops, it is important to suppress this effect (see e.g. Murri et al., 1998). Internal ply drops are less severe than external ones, i.e. with lay-up with a shielding continuous ply on the surface, since the stress concentration from interior ply drops will be substantially reduced (Chan, 1992, Cairns et al., 1999). Specimen design can be made with finite element analysis, in order to save costs and narrow down lay-up alternatives (Varughese and Mukherjee, 1997).

In three-point bending of short specimens, a measure of interlaminar shear strength may be obtained, which interpreted as the interfacial efficiency between the fibres and the matrix in unidirectional composites (Harvey et al., 1987). Miyano et al. (1995) performed three-point bending in fatigue of carbon fibre-reinforced plastics, and noticed in-plane shear failure which was particularly sensitive to cyclic loading at temperatures close to the glass transition of the polymer matrix. For thinner specimens, failure will instead occur in compression or tension at one of the surfaces. Dillon and Buggy (1995) noticed a progressive fatigue failure on the compressive side of the specimen for unidirectional carbon-fibre-reinforced plastics. To avoid indentation and crushing from the sole support on one side, four-point bending fixtures may be used.

2.6 Delamination tests

The interlaminar mode of fracture, or delamination, has aroused considerable attention during the last two decades. In a laminated composite subjected to service loads, it has become apparent that the delamination failure mode has the potential for being the major life limiting failure process (Wang, 1983; Chan,

1992; Carlsson and Byron Pipes, 1997). Delaminations are likely to be introduced in a composite laminate during processing of the lay-up as a result of contamination of the prepreg, leading to locally poor ply adhesion, or they may form locally in regions of high void content. More likely, delaminations may occur from post-fabrication handling or service in the structure. It is recognised that a delamination represents a crack-like discontinuity between the plies and that it may propagate during application of mechanical and/or thermal loads. It is thus appropriate to approach the delamination phenomena using fracture mechanics, which has indeed evolved as a fruitful tool for materials selection and structural integrity involving delaminations. Fracture mechanics of composites is mostly based on the strain energy release rate. There are many fracture tests for measuring the static and fatigue interlaminar fracture toughness, typically in terms of critical energy release rate, G_c, and crack growth rate versus the range of the energy release rate, da/dN-ΔG, respectively.

The susceptibility of composite structures to initiation and growth of delaminations must therefore be considered already in the design phase. Prediction of initiation and growth of delaminations is, however, complicated and the success of the predictions relies on accurate interlaminar toughness data for the material under both static and fatigue loading and at different environmental conditions. Furthermore, due to variations in mode conditions along delamination edges, toughness data must comprise pure mode I and mode II toughness as well as mixed-mode ones, ideally for different ply orientations at the delaminated interface. Delamination growth has shown to be significantly affected by its location within the ply stacking sequence of the composite laminate (Greenhalgh and Singh, 1999). The width of delamination specimens may in some cases influence the measured energy release rates (Robinson and Song, 1992).

Several different test methods have been proposed in the past for measuring the interlaminar toughness under various loading conditions. Today, interlaminar toughness in pure mode I is generally measured by the double cantilever beam test (DCB), whereas in mode II the end notched flexure test (ENF) is used. Delamination growth in structures is, however, rarely taking place in pure mode I or pure mode II and mixed-mode measurements are therefore important. A frequently used test method for mixed-mode conditions is the mixed-mode bending test (MMB), suggested by Reeder and Crews (1990). The MMB method has the advantage of using specimens similar to the DCB and ENF specimens, and thereby allows for a high degree of consistency in the interlaminar toughness measurements. For descriptions of the interlaminar toughness test methods, see Juntti et al. (1998), and for a comparison of the different kinds of tests, see Ireman et al. (1996) or Olsson et al. (1996). A promising complementary technique to mixed-mode delamination investigations is fractography, in which the mechanisms of mode I and mode II crack propagation can be distinguished (see e.g. Hiley, 1999).

Progressive crack growth, such as delamination, is generally characterised by a Paris law where the crack growth rate is expressed by a power law in terms of the maximum value or range of the energy release rate, viz.

$$\frac{da}{dN} = C(\Delta G)^m,$$
<div align="right">(1)</div>

where a is the delamination length, N is the number of cycles, ΔG is the energy release rate range. The constants C and m characterise the crack propagation rate, and are related to the intercept and slope in a Paris plot. For composite materials, the exponent m turns out to be an order of magnitude larger than most

metallic materials (Sjögren and Asp, 2001; Schön et al., 2000). The predicted propagation rate is therefore very sensitive in the accuracy of the estimated energy release rates. A design approach based on practical threshold values of ΔG is therefore preferable in many cases.
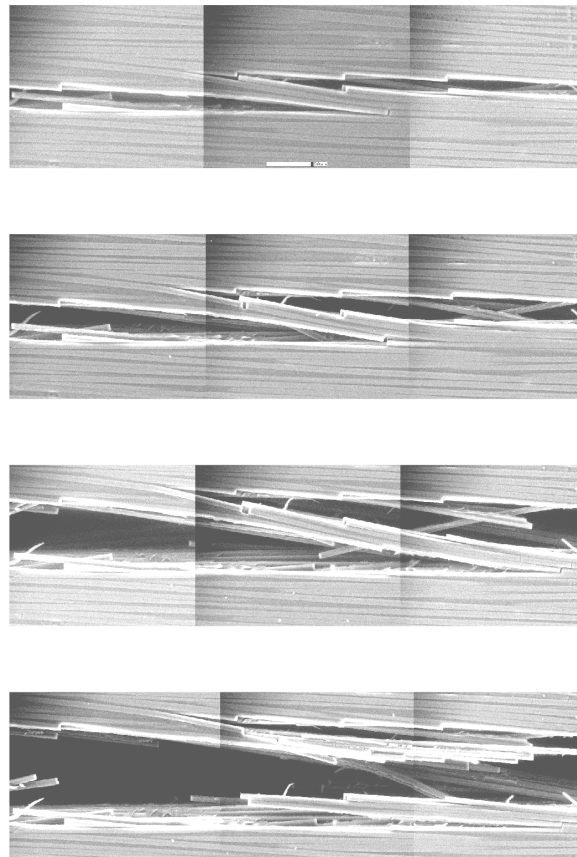
Figure 3. Micrographs of cross-over fibre bridging in delamination growth in a unidirectional carbon-fibre composite (Sørensen and Jacobsen, 1998b). The scale in the first micrograph indicates 50 µm.

Fibre cross-over bridging might occur in delamination growth. Under static loading in mode I in a DCB test, the micromechanisms of fibre bridging has been clarified (Spearing and Evans, 1992). In Figure 3, a series of micrographs is presented, where cross-over beam-like bridging takes place in DCB delamination test. On increased loading, the bridging fibre bundle beam eventually fails. This type of bridging results can be described by a bridging law, that relates the effective cohesive traction on the crack surfaces to the local crack opening displacement. Sørensen and Jacobsen (1998b, 2000) have shown that the R-curve may not be considered as material behaviour in the case of large scale bridging. In fact, the beam thickness of the DCB specimen influences the energy release rate prior to steady state. Erroneous design may therefore result if the R-curve is characterised for one specimen geometry and then applied to a structure with a different geometry. This can be circumvented by applying pure moment to the DCB specimens, and analysing crack growth in terms of the bridging law. The effect has not yet been studied under fatigue conditions, despite its practical implications.

2.7 Complementary non-destructive characterisation

Macroscopic mechanical characterisation, such as conventional stress-life data, may convey some useful information on how to rank different materials in terms of fatigue resistance. To get forward to conceive mechanism-based models, we must invoke an arsenal of non-destructive characterisation technique to continuously monitor the evolution of damage caused by the operative fatigue mechanisms. Destructive measurements such as residual strength measurements and penetrant-enhanced x-ray radiography will not give much information on how the damage grows with the number of applied load cycles since the material is fully or partially demolished when tested from damage due to chemical attack. Of special interest are instead non-destructive techniques that can be used in situ during the course of fatigue testing.

Acoustic emission is a result of stress waves emanating from zones where energy is released at initiation of certain damage mechanisms. The most energetic ones, typically transverse cracking, fibre breakage and delaminations, are readily picked up by piezoelectric sensors. The electric signal from these sensors can be correlated and identified with a particular failure mechanism. This characterisation has to be repeated for every new material and stacking sequence. Conventional acoustic emission measurements in fatigue consist of characterisation of a number of events at thresholded energy levels and at most a qualitative link to the types of damage sources (Bhat and Murthy, 1993). Recently, attempts have been made to quantitatively link the acoustic emission signals with the specific type of damage, and obtaining measures of both the position and the magnitude of the emitting crack in the matrix in transverse plies and in the fibres (Åberg and Gudmundson, 2000). The signal analysis of the transient measures can with advantage make use of suitable wavelets, where expected types of signals can be located both in time and space simultaneously. This method has been employed by Kamala et al. (2001) in fatigue experiments of carbon fibre reinforced composites. To fully use the potential of acoustic emission techniques to characterise the accumulation of damage in fatigue of composites, much more experimental evidence is necessary to conceive models that would work in a realistic setting.

Stiffness monitoring is a relatively easy method, since it only requires a continuous computer log of the change in strain from an extensometer during fatigue. With the addition of transversely mounted extensometers, the major Poisson ratio can also be logged together with the axial Young modulus. Stiffness monitoring is useful for applications with design situations where the failure criterion can be formulated as a decrease of the Young's modulus below a certain critical value (Brøndsted et al., 1996; Philippidis and Vassilopoulos, 1999). Moreover, if the stress-strain relation is measured continuously over individual cycles, the hysteresis loss can be obtained with the adherent loss factor responsible for damage accumulation and adiabatic heating (Osiroff and Stinchcomb, 1992). In laminates containing a substantial amount of off-axis plies, the reduction in axial stiffness can be directly correlated to the subcritical crack density of the plies with near transverse fibre orientation (Jamison et al., 1984).

Intermittent surface replication during fatigue is a useful tool to characterise the superficial microscopic damage (see e.g. Gamstedt, 1999a). A polymeric tape, usually made of cellulose acetate, is turned malleable by soaking it in an acetone based solution. It is thereafter pressed onto the surface of the composite. As the acetone softened tape flows into the surface and upon drying, a high fidelity impression of the surface topography on the micrometer scale is

obtained. The tape can then be covered a sputtered layer of carbon and film for scanning electron microscopy. This procedure is repeatedly done at different stages of the fatigue life in order to study the sequential evolution of fatigue damage. Great care has to be taken to keep the sample and replica clean, void of any debris that would make the structures on the replica spoiled or undetectable at high magnification. Naturally, damage processes within the bulk of the composite can only be indirectly deduced from what is observed to take place on the surface.

Ultrasonic scanning, on the other hand, readily gives a picture of damage entities within the material. The specimen is submerged in a water tank between a glass plate at the bottom and a transducer at the water surface of the tank. An ultrasonic pulse is emitted from the transducer, and the wave is partially reflected at the interfaces of damage within the composite material. The peak from the reflection at the glass plate is analysed and used as a measure of the attenuation in the material. A two-dimensional through-thickness image of the composite material is produced from the reflected acoustic signal. Delamination and larger voids are easily detected. A drawback is that the sample has to be exposed to water, which might result in chemical degradation, in particular for glass fibre based composites, where stress corrosion may occur (Caddock et al., 1990). Haque et al. (1993) investigated the sequential damage development under fatigue loading in cross-ply carbon fibre-epoxy laminates. The C-scans indicated incipient cracks almost immediately in the transverse direction, followed by splits in the longitudinal plies and local delaminations at the intersection points.

X-ray transmission radiography with an enhancing penetrant is widely used for the detection of larger damage in multidirectional laminates (see Simonds et al., 1989, for fatigue in laminates with drilled centre holes). Radiography takes everything into account that the x-ray beam encounters while going through the material and gives a plane projection of delaminations, pores, transverse cracks etc. Flaws perpendicular to the beam are therefore liable to escape detection. However, x-ray tomography confers a three-dimensional reconstruction of the interior of the material with a maintained resolution. When x-ray tomography is used, a collimated beam goes through the tested material and is received by an array of detectors on the opposite side. The object is gradually rotated through an angle of 180°. The attenuation of the intensity of the x-ray beam is measured and acquired by a computer which reconstructs the interior structure of the sample. The procedure is time consuming since the sample must be fully scanned for each increment in the rotation angle. However, improvements in resolution and scanning time are to be expected in the future. Considerable activities are undertaken in the development of this method for medical purposes. For x-ray radiography and tomography of composite materials, a contrast enhancing penetrant may be of benefit. Due to the small differences in density between the composite constituents and damage cavities within the material, a high-density penetrant like zinc iodide would render the cavities more visible provided they are connected to the surface and can be filled. Since a penetrant would alter the composition and mechanical properties of the composite, it cannot as a rule be used intermittently during fatigue. If the presence of the penetrant influences further degradation, the method is not fulla a non-destructive technique. X-ray tomography has been used to study the development of fatigue damage in notched quasi-isotropic laminates (Bathias and Bagnasso, 1992; Eriksson et al., 1997). Since the resolution of the medical tomograph was in the order of 1 mm, the smallest crack entities detected were delaminations and transverse cracks.

Laser Raman spectroscopy (LRS) has gained increased use during the last decade in experimental analysis of composite micromechanics. The underlying

principle is that the strain dependency of the frequency of the atomic vibrations in a crystalline material can be probed with LRS. Therefore, by loading a composite, the local strains in the visible fibres can be determined from the resulting frequency shift of the Raman spectrum. The fibres should have a crystalline microstructure, i.e. aramid and most carbon fibres are well suited, but amorphous glass fibres can not be directly analysed. The turbostratic nanostructure of the graphite layers in the carbon fibres is sufficiently crystalline to give a decent Raman spectrum with distinct peaks. This is the case in particular for high-modulus carbon fibres, where the molecular order is high. Rigged with an optical microscope, the focused laser beam can be directed onto a spot on one of the superficial fibres (see Figure 4 for principles). The strain profile can be determined by scanning along the length of the fibre. LRS has been successfully used to measure the fibre stress or strain in single fibre/epoxy coupons at a resolution of 1-2 μm. From the stress profiles of adjacent fibres, it is possible to infer the type of damage present between the fibres, e.g. debonding, matrix yielding or cracking. Since it is an optical technique, it does not require physical contact with the specimen and can therefore be used remotely. For example, Melanitis and Galiotis (1993) have used LRS to quantify the interfacial properties of a single fibre carbon/epoxy composite under fragmentation. A model composite with a monolayer of equidistant carbon fibres in an epoxy matrix was analysed, from which the interfacial shear stresses were deduced by van den Heuvel et al. (1997). Bennett and Young (1997) investigated the influence of a fibre-bridged crack on the fibre stresses in an aramid fibre/epoxy monolayer microcomposite. Marston et al. (1997) extended the technique for application on high-volume-fraction carbon fibre/epoxy composites.
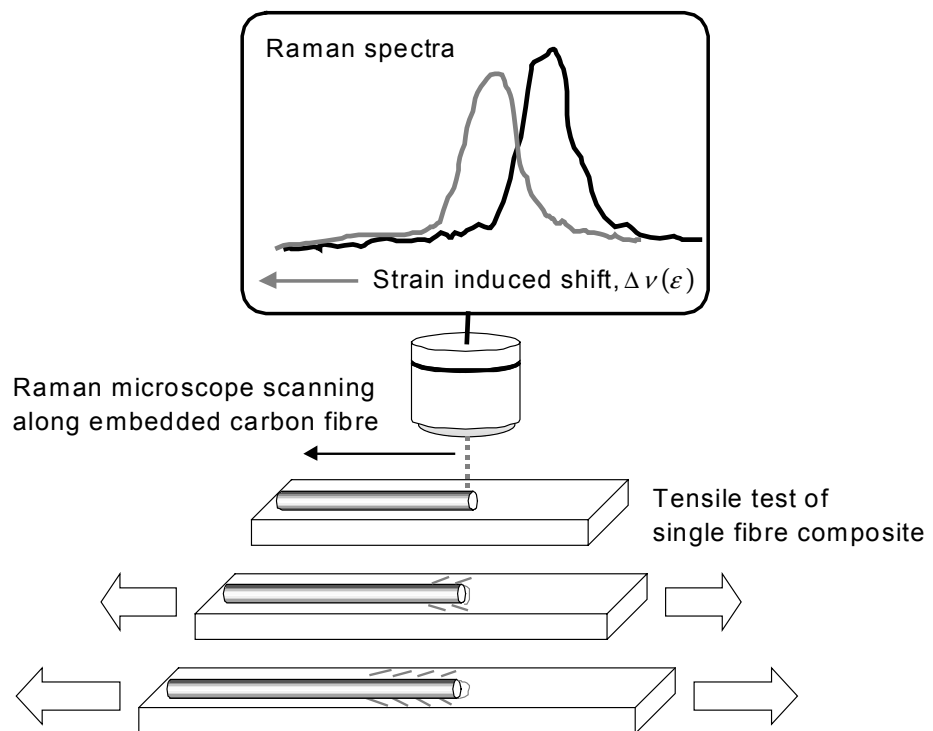

Figure 4. Schematic illustration of a mechanical test of a model micro-composite with laser-Raman spectroscopy

2.8 Post-mortem analyses

The most widely used post-mortem analysis method is fractography, either in a scanning electron microscope (SEM) or by conventional optical microscopy. If the damage processes have mainly occurred on the ply level or at larger scales, an optical microscope is likely to be sufficient, provided the depth-of-field is suitable at the given magnification. If smaller details are demanded, electron microscopes are a better option.

The objective of fractography in this framework is to trace back to the origin of structural failure, and identify the structural or material property that have incited the accumulation of damage leading to catastrophic failure. In this manner, sensible suggestions can be made of how to improve the material or how to improve the structure to subdue the deleterious damage mechanisms. One problem of fractography can be attributed to the heterogeneous microstructure with mixed brittle and ductile phases in composite materials, which makes it difficult to distinguish damage caused by fatigue or the catastrophic static failure at the end of the fatigue life. In composites, the brittle phase constitutes either the 0° or 90° plies on the laminate level, or the fibres on the microstructural level. The ductile phase corresponds to the off-axis plies where large shear deformation may take place, or to the polymer matrix in the microstructure. As cracks and other types of damage propagate during fatigue, they frequently cover both phases, and it requires considerable experience to interpret fractographic micrographs to figure out the crack propagation directions and rates in terms of e.g. striations, matrix rollers, shear cusps and other fractographic features (see e.g. Bascom and Gweon, 1989).

The typical process is outlined below. Firstly, the fracture surface is sputtered with gold or evaporised with carbon, and preferably analysed by scanning electron microscopy even though standard optical microscopy may be sufficient for large types of damage. If possible, the initiation region is first located, and then carefully examined for any indications of which damage processes that have been active. Since longitudinal carbon fibre composites are relatively brittle, a small amount of contiguous fibre breaks is sufficient to make the specimen fail catastrophically even at moderate stresses. As mentioned previously, locating the initiation spot is therefore an arduous task, which is further aggravated by the extensive splitting at final failure (Purslow, 1981). The elevated strain energy is suddenly released at failure, and the subsequent dynamic motion causes a multitude of longitudinal splits, which make the fracture surface uneven and difficult to analyse. However, fractography has successfully been used in examining compressive fatigue failure in three point bending (Dillon and Buggy, 1995). Matrix cracks are generally more useful in fractographic analysis to track down the original source of failure (Purslow, 1986). Fractography of crack surfaces formed by fatigue is in general more difficult to analyse than those formed from quasi-static loading, since a wider variety of operative damage mechanisms have been active throughout the course of the fatigue lifetime.

3 Fatigue damage mechanisms

3.1 Tensile fatigue of longitudinal laminates

Due to the complexity that arise from the influence of applied loading mode and laminate lay-up on the damage mechanisms, some simplifications are required as a first step in a mechanism-based approach in order to narrow down the problem and to identify the dominating and critical microstructural aspects. If all possible parameters were taken into consideration, e.g. a general laminate lay-up subjected to variable multiaxial tension-compression loading in severe service environments, the micromechanical scenario becomes too complicated to be analysed with a more exact physical approach. We would then be forced and compelled to a conventional macroscopic procedure, which does not allow for any extrapolative predictions or indicate how the structure of the material can be improved. The simplifications should be made judiciously to maintain relevance in view of the intended applications. If we consider the stacking sequence of a general multidirectional laminate, it is the longitudinal ply that is the critical element, since it is generally the main load-carrying constituent, and is the last one to fail. If we were constrained only to focus on damage mechanisms in one class of composites, it would have to be those in unidirectional composites with continuous longitudinal fibres, i.e. with the fibres in the same direction as the load direction in uniaxial tension. In design of composite components, one generally has an idea of the expected stress state in the structure. It is then sensible to arrange a considerable amount of the constituent plies with a fibre direction parallel to the direction of the largest principal load. Numerous investigations have shown that the longitudinal plies dominate the fatigue-life behaviour of general multidirectional laminates, when interpreted in terms of global applied strain (Talreja, 1993; Dickson et al., 1989; Andersen et al, 1996). Concerning the loading mode, tensile fatigue with a stress ratio of $R = \sigma_{min}/\sigma_{max} = 0.1$ is first considered. The effect of the simple loading modes must first be clarified, before moving on to more complex loading modes, such as variable multiaxial tension-compression loading.

In Figure 5, a so-called fatigue-life diagram is presented, which makes the interpretation of fatigue failures in tensile fatigue of composites more tractable (see e.g. Talreja, 1981). The *S-N* data are replotted with maximum strain attained in the first load cycle along the ordinate and the logarithm of the number of load cycles to failure along the abscissa. The axial strain level in a longitudinal composite is globally the same for the fibres and the matrix. The strain description has also an advantage for multidirectional laminates, since all plies are globally subjected to the same strain for in-plane loading, i.e. no significant bending. The static strain to failure is in practice controlled by the strain to failure of the fibres. Furthermore, the maximum initial strain in conventional load-controlled fatigue is linked to the damage state reached in the very first load cycle, and there is reason to believe that any progression of damage during the course of fatigue is determined from this initial damage state. A baseline fatigue-life diagram, as the one presented in Figure 5, pertains to a unidirectional composite loaded in the fibre direction, and consists of three distinct regions, each associated to a different set of operative damage mechanisms. For notational simplicity, these regions are enumerated as Regions I, II and III. Fatigue life data for composite materials generally present themselves in these regions with discontinuities in slope in an *S-N* plot.

In Region I, the strain level generally coincides with the scatter band of the static strain to failure, since virtually all the load is carried by the fibres in

conventional polymer-matrix composite materials. The operative mechanisms are similar to those in static failure, and can be regarded as chaotic in the sense that it is next to impossible to predict the lifetime distribution due to the sensitivity to the slightest variation in the state of damage developed in the first loading cycle. Static fracture is a stochastic process controlled by the fibre strength distribution and the stress redistribution from fibre breaks. From a design point of view, the fatigue behaviour in Region I is not of central interest, since the lifetimes are short and strain levels can be analysed in terms of static strains to failure.
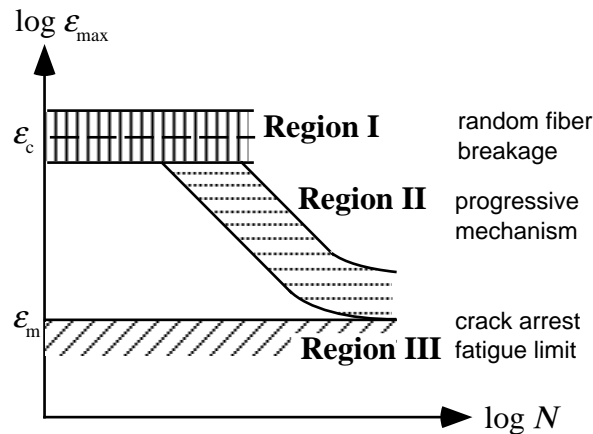

Figure 5. Fatigue-life diagram used for mechanistic interpretation of S-N *data.*

Region II is also called the progressive region. Candidate damage mechanisms responsible for the downward slope of this region are progressive fibre bridged cracking (Gamstedt and Östlund, 2001a) and progressive debond propagation (Dharan, 1975; Lorenzo and Hahn, 1986; Gamstedt, 2000b). Fatigue sensitivity is generally characterised by the slope in a fatigue-life diagram or in equivalent plots. Such a slope implies that there has to be progressive damage mechanisms, i.e. cycle-dependent growth of cracks, and that the progression rate depends on the applied load level. The progressivity is intrinsic to Region II in a fatigue-life diagram, in contrast to the horizontal Region I, where damage growth is cycle independent. Progressive fibre-bridged cracking and/or debonding must also give rise to fibre failure, because all fibres must be broken at least in one place for failure to occur. With higher demands on lighter and more slender structures, especially in aerospace applications, design for finite fatigue lifetime has become a reality. Therefore, the damage mechanisms in the sloping Region II and controlling microstructural parameters have become subject to increased attention, since this region constitutes the part of finite lifetime, which is at least realistic for design in its high cycle parts.

It is more common to design against the fatigue limit, i.e. structure dimensions are chosen so that strains or stresses above the fatigue limit are not allowed in the composite. Below the notional fatigue limit, no failure is expected within an indefinitely long time or a certain chosen high number of cycles, say 10^8 or 10^9 cycles. This region is called Region III. The strain levels here are so small that damage does not develop at all, and the material behaves perfectly elastic in cyclic loading. A more plausible scenario is that damage does form in this region since the fibres have a tailed distribution in strength, but that any progression of damage becomes constrained and arrested by the inherent heterogeneous microstructure of the composite. Alternatively, the applied strain level may be so low that the driving force for propagation from

static damage sites does not reach above its threshold level. Knowledge about the operative damage mechanisms in Region III is useful in the sense that it provides us with information of how damage propagation can be retarded and arrested. Conservative design can then be made of materials that will last practically interminably, unless other damage mechanisms come into effect at very high cycle fatigue, e.g. stress corrosion in glass-fibre composites (Price and Hull, 1983) which is a diffusion controlled process.

In the following, attention will be focused on fatigue damage mechanisms in Region II, since this region will become of vital importance when finite life design will become more frequent in practice. Several studies have shown that progressive fibre-matrix debonding is the principal mechanism in fatigue degradation in Region II. Starting with static properties, the fibre-matrix interface has shown to be of primary influence in strength of carbon-fibre reinforced plastics, in particular for lay-ups where deviatoric strains prevail (Drzal and Madhukar, 1993). The results showed considerably lower strengths for loading modes involving oblique fibre orientation with concomitant shear stresses. The loading case that showed the least influence of interfacial adhesion was unidirectional longitudinal tension, where a difference in static strength was barely noticeable. This is not the case for fatigue loading, where the interface efficiency has shown to give rise to a large difference in lifetime (Gamstedt and Talreja, 1999a; Gamstedt et al., 1999b). The reasons for this will be outlined below.

A generic feature for fibre reinforced plastics is that the brittle fibres are the first to fail, and they are surrounded by a more ductile polymer matrix. In static as well as fatigue loading, the first failure events that will occur are thus distributed fibre failures. Since the fibres have a distribution in strength, with a tail for low strengths that cannot be neglected, there will be a number of fibre breaks, even at very moderate loads. The local stress will change as the fibres break. In the vicinity of the fibre break a stress concentration is likely, which can activate further progression of damage. Debonding along the fibre-matrix interface will greatly influence the kinetics of this damage progression. Depending on the efficiency of the interface, we can envisage two different scenarios with a single fibre break, which are schematically depicted below in Figure 6.

(a) Fibre-bridged cracking   (b) Fibre breakage by debonding
                                                  and yielding
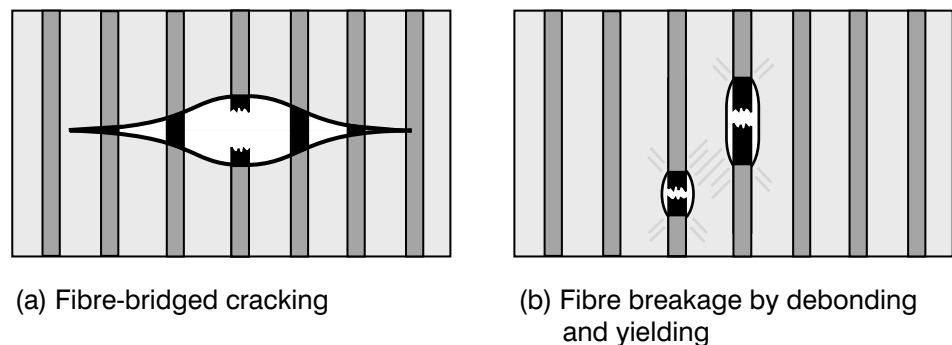
Figure 6. Crack propagation in 0° polymer-matrix composites from a single fibre break; (a) fibre-bridged cracking, and (b) successive fibre breakage by fibre-matrix debonding and matrix shear yielding.

If the fibre-matrix interface is strong, matrix cracks will grow from the distributed fibre breaks. These matrix cracks will grow perpendicularly to the fibre and load direction, and as the cracks reach the neighbouring fibres, the cracks will propagate around them. A fibre-bridged crack will form in this way.

For the crack to open, there has to be some slippage between the bridging fibres and the surrounding matrix directly above and below the crack in the bulk composite. This slippage is mostly caused by limited fibre-matrix debonding, or by local shear yielding of the matrix. The slippage influences the cohesive action brought about by the bridging fibres. The resulting effective crack surface traction tries to close the crack, and thereby reduces the driving force for crack propagation. The growth rate is therefore indirectly influence by the fibre-matrix damage process in the crack wake. A plausible failure criterion would be the rupture of the furthest fibre in the wake. The stress in this fibre would also be strongly influenced by the amount of fibre-matrix slippage. This criterion is supported by modelling work (Begley and McMeeking, 1995), and experimental observations (Gamstedt and Östlund, 2001a), which show no indication of stable fibre breakage in a fibre bridged crack.

However, fibre bridged cracking, as described above, is relatively rarely reported for longitudinal fibre reinforced plastics (see e.g. Ritchie et al., 1989; Luke et al., 1996; Botsis et al., 1995; Shih and Ebert, 1987). This is presumably because of the rather high volume fraction of fibres, high bond strength and ductility of the matrix. A more common type of damage sequence is governed by far-reaching fibre-matrix debonding with successive breakage of fibres. As an example, a series of surface replicas of a carbon fibre/epoxy specimen subjected to fatigue has been presented by Gamstedt and Talreja (1999a). The fibre-matrix debond process is shown to be a progressive mechanism, that initiates from fibre breaks, and grows in mode II in the same direction as that of the fibres and applied load. Curtis (1991) has observed the same phenomenon from small drilled holes in unidirectional carbon-fibre-reinforced plastics. It can be envisaged that the stress profile in the neighbouring fibres will change in a non-monotonic manner as the debond propagates in fatigue. Eventually, it is likely that a weak segment in a neighbouring fibre will be locally over-stressed, and thereby rupture. This would give further impetus to debond growth in a self-escalating manner. These ideas can be outlined in a schematic way.

$N=1$ $N=N_0$ $N=N_1$ $N=N_2$ $N=N_f$

Distributed fibre due to their strength distribution   Debond growth from fibre breaks, stress redistribution in load carrying fibre segments   Further debond growth, additional breaks of over-loaded fibres   Progressive debonding and fibre breakage, crack coalescence   Failure of specimen by separation
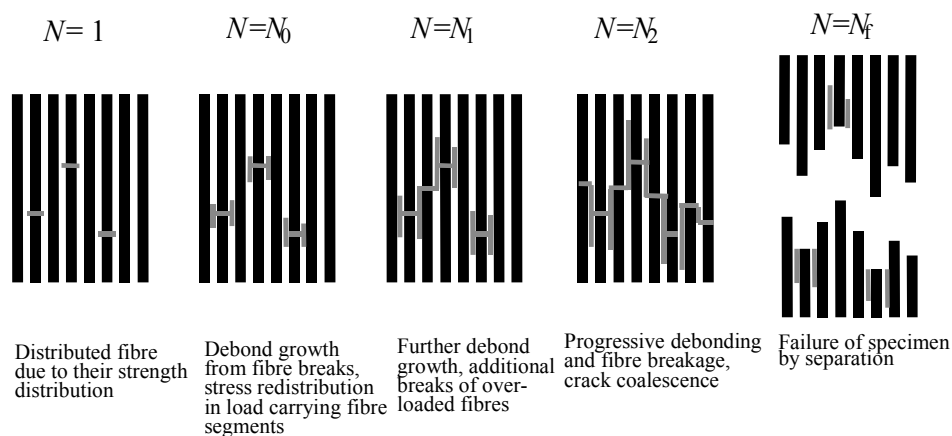
Figure 7. Schematic illustration of fatigue damage accumulation and failure by debonding in a unidirectional composite loaded in tensile fatigue in the fibre direction.

The debonds seem to initiate from the distributed fibre breaks. Since the brittle fibres have a distribution of strength, there will be a multitude of distributed fibre breaks in the volume of the material after the first application of load, even though this load would only induce a very moderate stress well below the

ultimate strength of the composite (cf. Figure 7, $N = 1$). As the debonds grow, the stress profiles in the neighbouring fibres continually change ($N = N_0$). As suggested above, some segments will be over-stressed, and because of the strength distribution along the fibres, weak parts will fail ($N = N_1$). This will in turn give rise to further debonding ($N = N_2$), etc. Eventually, the entire specimen will break when all fibre breaks have linked up through debonding ($N = N_f$). This type of damage evolution is spatially distributed, and results in a global and uniform degradation of the material until just before failure. A more brittle behaviour manifests itself in e.g. carbon fibre/epoxy with its strong interface, which gives rise to localised damage where the most severe crack is the sole cause of failure. Fibre-bridged cracking as outlined above may even occur in polymer-matrix composites with efficient and fatigue resistant interfacial stress transfer. Materials with a stronger interface tend to show brittle-like planar fracture surfaces, where as weak-interface materials exhibit brush-like failures with protruding fibres, which is illustrated in Figure 8.
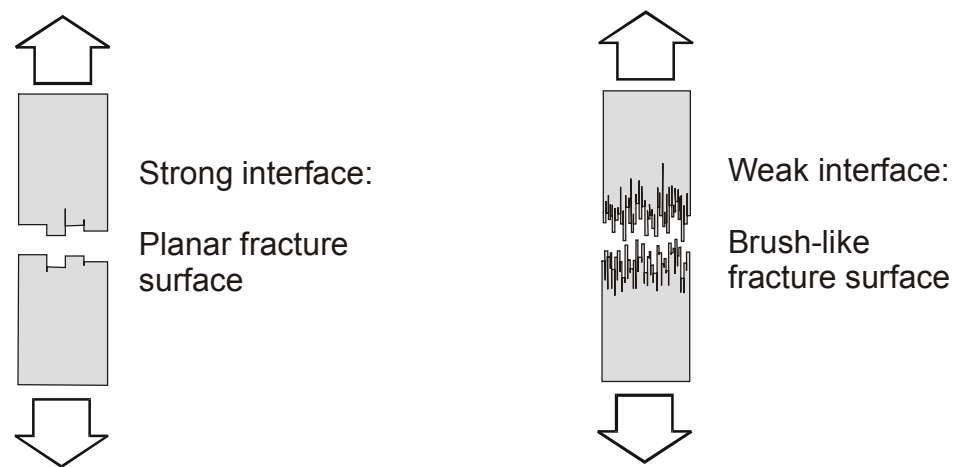

Figure 8. Schematic illustration of the influence of fibre-matrix interfacial strength on tensile failure mode.

Despite a large difference in constituent properties, the influence of fibre-matrix cohesion on damage accumulation and fatigue behaviour shows similar traits for unidirectional carbon-fibre reinforced plastics (Gamstedt and Talreja, 1999a), for glass-fibre reinforced plastics (Gamstedt et al., 1999b) and for glass fibre/carbon fibre hybrid composites (Gamstedt and Brøndsted, 1999d). The mechanisms are not material specific. A comparison of the scenarios with transverse damage growth and longitudinal damage growth in different composite materials has been made by Gamstedt (2000a), with special regard to the influence of the fibre-matrix interfacial mechanisms. In fact, the two different mechanisms can appear in one and the same material, depending on which load amplitude that has been applied. At low strain amplitudes close to the fatigue limit, the debonding scenario has partially presented itself in a carbon fibre/epoxy material as a 'sub-mechanism' in fibre bridged cracking. Transverse bridged cracks were then effectively arrested by fibre-matrix debonding. The debonds grew to a limited extent, and then stopped, after which no further development of damage was detected. In this case, the debonding acts as an arrest mechanism to progressive matrix cracking. The role of debonding for this material is schematically presented in Figure 9.

A general remark, which is valid for both cases of fibre-bridged cracking and of successive debonding with fibre breakage, is that damage sites like these are principally unavoidable in fatigue of longitudinal composites. The considered composites consist of a brittle and strong phase – the fibres – and a ductile and weaker phase – the polymer matrix – that transfers stresses between the load carrying fibres. Even though the thin fibres are ever so strong in the ideal case, they are brittle and hence flaw sensitive. A stochastic occurrence of flaws from processing considerably reduces the strength of long fibres. The inherent strength can only be retained if the fibres are moulded into a matrix to form a composite. In that case, the material can sustain considerable loads even though a large amount of the fibres are broken, since the matrix effectively transfers load from broken fibres to intact ones by shear stresses. As mentioned before, there will inevitably be a handful of broken fibres, due to the distribution of flaws in the brittle fibres. Damage progression from these flaws is also inevitable unless the applied stress or strain is maintained well below the fatigue limit. The microstructural heterogeneity thus makes the material intrinsically prone to damage accumulation. Luckily, with a sensible composition and a suitable fibre-matrix interface, these materials can be made significantly tolerant to damage. The ability to transfer stress concentration from damaged volumes to undamaged material volumes gives the composite its extraordinary properties.
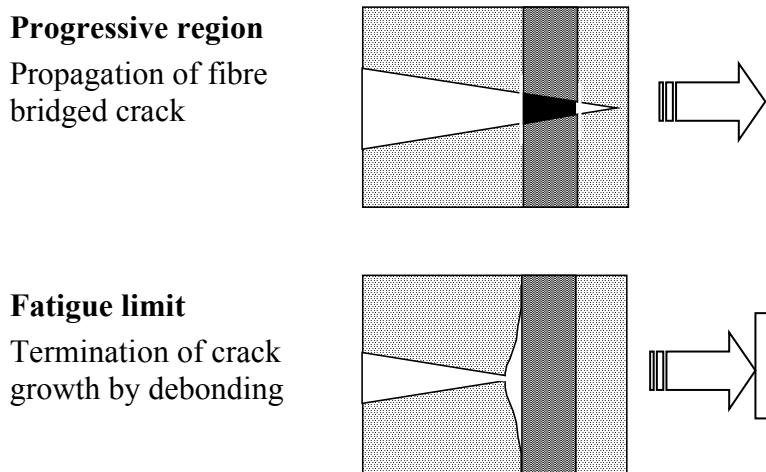
Progressive region

Propagation of fibre bridged crack

Fatigue limit

Termination of crack growth by debonding

Figure 9. Illustration of the role of limited debonding as an arrest mechanism in loading at the fatigue limit.

It is clear from the results presented in this survey, and from references herein, that a strong fibre-matrix interface, which does not allow for any widespread progressive debonding, would as a rule result in improved fatigue resistance, in the sense that the *S-N* curve would become relatively horizontal. This means that the material either fails at the first application of load, or it sustains an interminable number of repeated equal load cycles. On the other hand, an improved interface would also mean embrittlement, which implies above all an increased sensitivity to flaws and other sources of stress concentration in conditions of static loading or impact. Even static strength is known to decrease if the stress transfer is too efficient (Marston and Neumeister, 1999). For optimal materials design aiming for maximum finite life at a certain stress amplitude above the notional fatigue limit, one would therefore expect that there would be an optimum in fatigue debond growth resistance.

3.2 Tensile fatigue of multidirectional laminates

The most important ply in a multidirectional laminate is the longitudinal ply. On axial loading, this is the ply that has the largest strain to failure, is stiffest, and has the longest fatigue life for a certain applied strain level. In many cases it seem to control the fatigue life of multidirectional laminates. In Figure 10, a fatigue-life diagram is presented in which the *S-N* data for unidirectional longitudinal as well as for various multidirectional laminates are plotted. At least for high-cycle fatigue close to the fatigue limit, data from different laminates fall on the same scatter band. However, if the off-axis plies contribute to a substantial part of the laminate, the damage that develops in these plies will affect the damage evolution in the longitudinal ply. If a rough estimate in tensile fatigue life in terms of peak strains for a general lay-up, the *S-N* data for a unidirectional longitudinal laminate of the same material can be used. For more accurate life predictions and estimation of stiffness degradation due to damage accumulation, the mechanisms specific to the multidirectional laminate should definitely be addressed. This degradation can either be accounted for by an empirical power law (see e.g. Halverson et al., 1996), or by more detailed modelling of the mechanisms (e.g. Akshantala and Talreja, 1998). The reason for incorporating various fibre directions in multidirectional laminates, and mechanisms of damage accumulation in the off-axis plies and their interfaces will be discussed in this section.
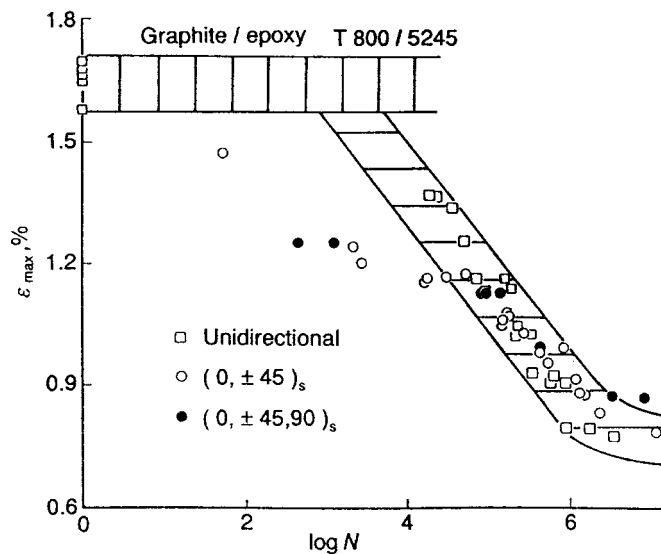

Figure 10. Fatigue-life diagram of [0]$_T$, [0, ±45]$_S$ and [0, ±45, 90]$_S$ carbon fibre/epoxy laminates in tensile fatigue (R = 0.1) from Talreja (1993).

Composite laminates are usually composed of individual plies with different fibre orientations. The transverse plies and symmetric plies with oblique angles are almost ubiquitous in laminates used in applications, since stresses are rarely uniaxial in load carrying structures. A pure unidirectional longitudinal lay-up would be very sensitive to splitting, and to avoid this loss of structural integrity, a small amount of transversely oriented fibres is almost always added, even in cases where a uniaxial stress scenario is anticipated. The transverse plies show the lowest strength and strain to failure. It is usually in these plies where damage first accumulates, from which it can progress into more critical plies. In some applications, no damage may be tolerated at all, e.g. in pressure vessels,

where cracking in the transverse plies would result in leakage. The transverse ply is then the critical element, and the damage mechanisms in this ply should be investigated in detail. This section concerns the influence of the fibre-matrix interface on damage initiation and propagation in fatigue of transverse plies. An off-axis fibre orientation may behave similarly, although with a contribution from matrix shear deformation. It is clear from Figure 11 that the interface plays an important role in the formation of cracks in transverse loading. As in the case of longitudinal composites, the importance of the fibre-matrix mechanisms in fatigue is even greater (Harrison and Bader, 1983; Gamstedt and Sjögren; 1999c). Incidentally, the influence of the fibre-matrix debond process on fatigue damage accumulation and failure is particularly noteworthy in tension-compression loading, since the inhomogeneous microstructure allows debond propagation by crack opening during the compressive load excursions.
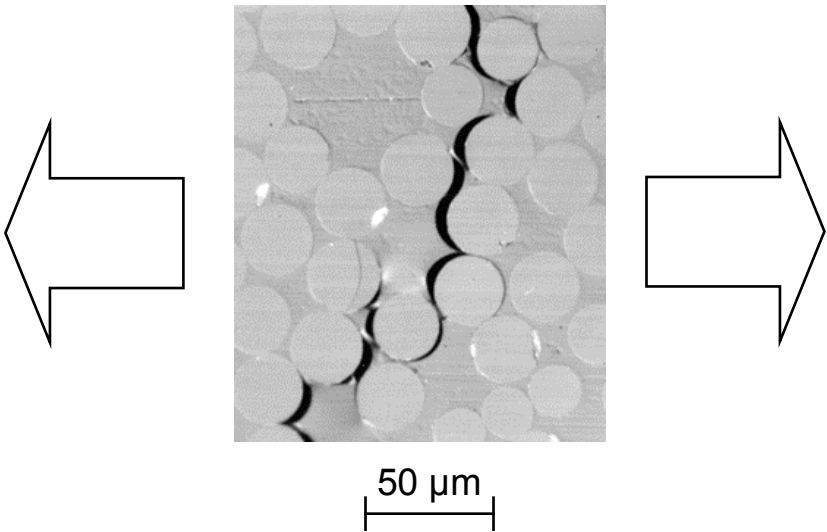

50 µm

Figure 11. Optical micrograph of a transverse crack forming from coalescence of fibre-matrix debonds in transverse loading.
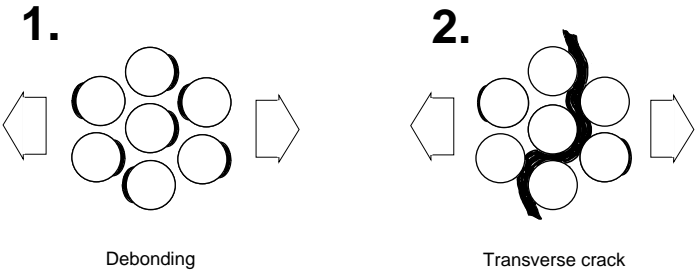
Figure 12. Schematic illustration of coalescence of fibre-matrix cracks into transverse cracks in plies with transverse fibre orientation.

In Figure 12, a schematic picture is found that depicts the formation of transverse cracks. In many cases, the interface is weaker than both the fibre and the matrix, meaning that cracking first takes place in the bimaterial interface. The difference in elastic properties promotes crack growth along the interface, both in static and fatigue loading. Eventually, debonds propagating around fibres will coalesce and form macroscopic transverse cracks spanning across the transverse ply, or would lead to ultimate failure by separation for a

unidirectional transverse composite. If the progressive fibre-matrix debond growth is suppressed, the lifetime can be extended. Fibre failure is not likely in unidirectional transverse composites, since the applied stress has an upper bound at the matrix strength, and the fibres would merely act as stress raisers. The stress concentration effect and residual thermal stresses imply that the fibre architecture or spatial distribution influences the transverse strength (Bulsara et al., 1999).

As mentioned previously, the damage that develops in different adjacent plies will mutually influence one another. For instance, fibre-matrix debonds may coalesce to form transverse cracks in transverse plies, which will lead to stress concentrations in the nearby on-axis plies. If we consider the simplest multidirectional laminate, i.e. a cross-ply laminate with a $[0_n, 90_m]_S$ lay-up, damage would accumulate as shown in Figure 13. The transverse cracks induced by debond coalescence multiply until saturation. The stress concentrations in the critical longitudinal plies will influence further damage progression and local degradation. Transverse cracking is a rather initiatory mechanism in comparison to e.g. progressive longitudinal debonding. Other progressive mechanisms inherent to multidirectional laminates have been identified, and will be outlined in the following.

We have noticed that fatigue cracks tend to grow at interfaces between media of different elastic properties. Microscopically, this manifests itself in progressive debonding between fibres and matrix. At a mesoscale, there are also bimaterial interfaces that are prone to cracking. If the plies of different fibre orientation are separately homogenised to form effective ply moduli where the detailed fibrous microstructure is ignored, an interface between plies of different anisotropic stiffnesses presents itself. Delamination growth along this interface is particularly a nuisance in fatigue loading, since delamination have shown to be a progressive mechanism that is less active in static in-plane loading. If compressive load excursions are applied, out-of-plane buckling of delaminated plies may result in unstable crack growth and member failure.
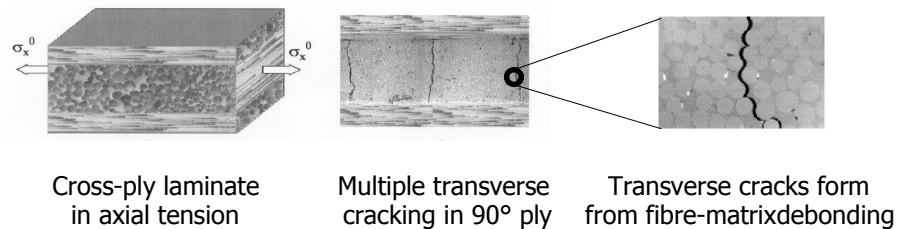

| Cross-ply laminate in axial tension | Multiple transverse cracking in 90° ply | Transverse cracks form from fibre-matrixdebonding |

Figure 13. Formation of transverse cracks in transverse plies from coalescence of fibre-matrix debonds (Joffe, 1999).

Progressive delamination has been observed to initiate from transverse crack tips, and as the delaminations grow the stress state in the transverse plies will change. In this way, delaminations will influence the continued transverse cracking, and vice versa. A typical delamination grown in fatigue along the 0°/90° interface from a transverse crack tip in a cross-ply laminate is shown in Figure 14. It can be noticed that the density of fibre breaks in the longitudinal ply is higher in the vicinity of the transverse crack and its delamination. With a sufficiently high density of fibre breaks, nearby fibres will fail unstably, and thereby cause complete failure of the specimen. The mutually influencing accumulation of transverse cracks in off-axis plies and delamination between plies, will continually change the local stress state in the critical element, in this

case the longitudinal ply. The stress history of the critical element dictates the lifetime of the composite structure.

It has been identified that the longitudinal ply is the critical element if present in a multidirectional lay-up. The micromechanisms in fatigue of this ply have also been identified, i.e. stochastic fibre breakage accompanied by fibre bridged cracking or progressive debonding. The rate of these mechanisms is controlled by the local stress state, which in turn is influenced by the damage that develops in the off-axis plies of the laminate, e.g. transverse cracking, shear matrix yielding or cracking, and interlaminar delamination. If the mechanisms on the ply level are analysed separately, it would be possible to estimate the local stress history of the longitudinal plies, and thereby predict the degradation in these plies until it fails (Akshantala and Talreja, 2000). The analysis of mechanisms in a multidirectional laminate can thus be divided into two different approaches with different size scales. 'Damage mechanics' deals with the ply-level mechanisms and ensuing degradation, which influence the local stress state of the constituent plies, and the rate of degradation can locally be described by 'micromechanics'. These disciplines are usually treated separately. The former one, damage mechanics, is useful to relate to macroscopic properties, such as residual stiffness. The latter one, micromechanics of the critical elements, is necessary for more accurate life prediction and materials tailoring. A link between the two may be established with a continuum model for determining mechanical response of composites with micro-scale damage. By use of a thermodynamics-based formulation of constitutive relationships with internal variables, the stress-strain-damage relationships can be derived. Specific expressions can be found for changes in the engineering elastic moduli in terms of the densities of damage entities for sparsely distributed damage (see e.g. Talreja, 1991).

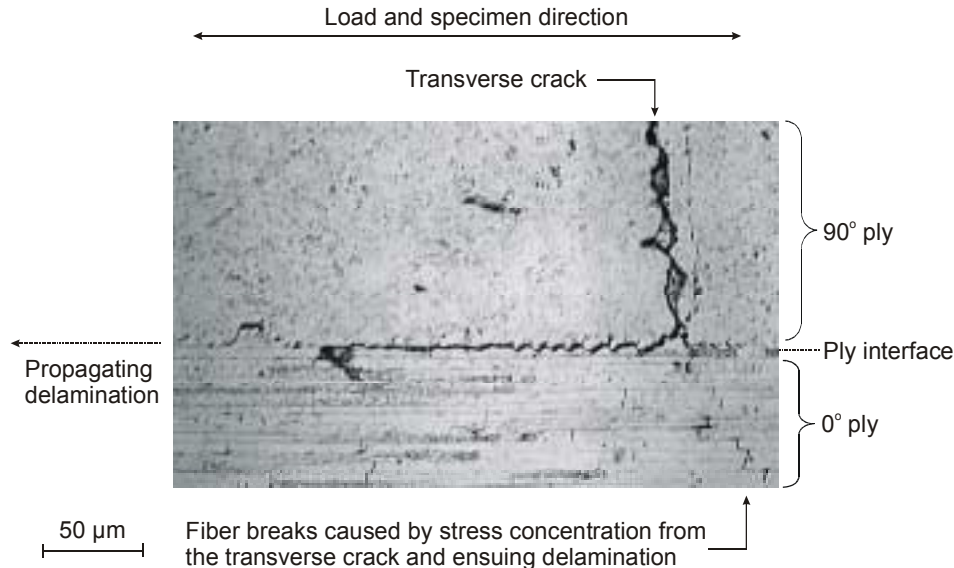*Figure 14. Replica section of an interlaminar fatigue delamination initiated from a transverse crack, and fibre breakage caused by arising stress concentrations.*

The simplest multidirectional laminate, the cross-ply configuration, shows some similarities with unidirectional longitudinal specimens, in the sense that they are composed of two phases; one is relatively brittle, i.e. the ensemble of fibres and

the transverse plies, and that progressive cracking tends to take place between this brittle phase and the other more ductile one, i.e. debonding and delamination, respectively. There is no coincidence that neither of the two materials complies with the Palmgren-Miner fatigue rule. Both tend to show shorter lives for a high-low load sequence compared with the opposite order. The similitude in variable amplitude loading has been shown in an experimental study of a carbon fibre/epoxy cross-ply material (Gamstedt and Sjögren, 2001b). In this case an abundant amount of transverse cracks formed if the high amplitude block was the first one, and in the second lower amplitude block frequent delaminations could peacefully propagate and lead to a significant overload of the main load-carrying constituent, which was the longitudinal plies. This led to premature failure, and unconservative estimates of the Palmgren-Miner rule. When the block load order was reversed, not much damage was formed in the first low amplitude load block, since there were not very many transverse cracks from which progressive delamination could grow. When the second high amplitude block set in, only some more transverse cracks appeared. The load-amplitude sensitive transition of the synergistic mechanisms can thus mechanistically explain the inability of the Palmgren-Miner rule to predict fatigue failure in composites, both on ply level for unidirectional longitudinal orientation and on the laminate level for cross-ply laminates. Non-linear generalisations of the summation rule of Palmgren-Miner is usually used to account for this kind of behaviour (see e.g. Marco and Starkey, 1954), but with the use of 'damage mechanics' can with advantage be used to describe this phenomenon more clearly and on a physical basis.

3.3 Compressive fatigue

Compressive fatigue is much more deleterious in composites than in apparently homogeneous materials such as metals and alloys (see e.g. Bathias, 1991). Delamination from defects and notches has been pointed out as one of the main sources of the compressive fatigue sensitivity. Cox et al. (1992) have investigated the damage mechanisms in compressive fatigue of composites with three-dimensional reinforcement. In almost all cases, it was found that the main failure event was kink band formation in the primary load bearing components, i.e. the axial tows. The characteristics of the reinforcement geometry that were observed to influence kink band formation was typically initial misalignment of the load bearing tows and intersections of load bearing tows and through-thickness reinforcing tows. Kink bands or fibre microbuckling emanating from defects have also been observed by Purslow and Potter (1984), and by Guynn and Bradley (1989). In general, the fatigue mechanisms are similar to those present in quasi-static loading, with the additional feature of its kinetics and a general tendency to grow along the fibre direction to larger extents. The mechanisms and mechanics of compressive composite failure in static loading have been presented by Budiansky and Fleck (1993), and Fleck (1997), with special emphasis on kink band formation and propagation and longitudinal splitting. A generalisation to fatigue loading has also been presented (Slaughter and Fleck, 1993), where the kinetics of kink band formation is controlled by Mroz plasticity in the matrix, and by the Armstrong-Frederick ratchetting law. However, no microscopic measurements of propagating damage on the micro-level have been correlated to the proposed models. A microscopic investigation of the operative fatigue damage mechanisms by Pruitt and Suresh (1992) in compressive loading of a unidirectional carbon fibre/epoxy composite is in support of the modelled mechanisms. A tendency of formation of two or more parallel cracks and less pronounced kink-bands was observed.

Due to the abrupt failure in compressive fatigue, and the difficulties to experimentally capture the failure events taking place prior to catastrophic failure, there are very few published results on the mechanisms, and the predictive models for property degradation and failure tend to be macroscopic and empirical in nature. A more thorough experimental characterisation concerning compressive fatigue would greatly benefit the understanding, and the possibilities to develop more accurate and useful models.

3.4 Tension-compression fatigue

Tension-compression fatigue is a common type of load wave form for service applications (e.g. in vibrating structures such as rotor blades for helicopters and wind turbines). The greater part of all laboratory fatigue testing is however performed under tension-tension cycling. The reason for this unbalance may be explained by difficulties in testing, such as buckling of specimens and lack of control of the stress state. Furthermore, compressive loading is likely to induce substantial stress concentrations in the vicinity of the grips, which gives rise to apparently large volume effects (Daniel and Hsiao, 1999).

For composite materials, tension-compression fatigue has shown to be more deleterious than tension-tension fatigue. The *S-N* curves plotted with the peak maximum stress or strain along the ordinate show significantly steeper slopes for load wave forms containing compressive load excursions. This has been shown to be the case for multidirectional laminates containing off-axis plies (see Rosenfeld and co-workers, 1978, 1981) and in particular for those containing transverse plies (Ryder and Walker, 1977; Gathercole et al., 1994; Nyman, 1996). Detrimental effects of compressive load excursions has also been observed for pure unidirectional transverse composites (Rotem and Nelson, 1989; El Kadi and Ellyin, 1994). The physical reason for the adverse effect of tension-compression loading of composite materials has not been clarified entirely. For multidirectional laminates or unidirectional laminates with distinct resin-rich inter-ply regions, delamination followed by out-of-plane ply buckling would certainly play an important role on the ply-level scale. On the scale of an individual fibre or a group of fibres, the underlying mechanics remains more unclear. If this could be clarified, further progress in the development of more predictive tools and constructive materials design is be anticipated.

For materials with a more homogeneous microstructure fatigue crack growth is characterised by self-similar propagation of single large cracks until the largest one of them achieves a critical size and catastrophic failure takes place. This propagation is primarily governed by crack opening under remote tensile loading. A compressive load excursion would close the crack tip and effectively result in negligible propagation of the crack in comparison with corresponding tensile load fluctuations. However, for a heterogeneous material like a composite that consists of phases with differing elastic properties, distributed microcracks of various sizes and shapes would form at an early stage of loading. The interfaces between the constituents would be particularly prone to cracking due to the inherent stress concentration and residual stresses from processing. Considering the cracks between media of dissimilar elastic properties, it is likely that some of these microcracks would open under global compression, especially those oriented along the load direction. In a multidirectional composite laminate with continuous fibre plies, cracks with an effective opening under compression could be incipient delamination on the ply level and large fibre-matrix debonds on the microscopic level. For notched composite specimens, it has been reported that cracks grow parallel or perpendicular to the

load direction depending on whether the load is on the compressive or tensile side, respectively (Stinchcomb and Reifsnider, 1979).

The formation of transverse cracks can be initiated by microcracks which grow under compressive loading. It is not far-fetched to suggest that this type of microcracks would be interfacial debonds with a fairly large circumferential size. Debonding is known to be the micromechanism incipient to transverse cracking. As the debonds at the interface between the fibres and matrix coalesce and grow unstably, a transverse crack forms (Harrison and Bader, 1983). Even if it is assumed that transverse cracking is initiated in the matrix through yielding or cavitation, it has been shown that this will take place infinitesimally close to the fibre-matrix interface for uniaxial transverse loading (Asp et al., 1996). This crack initiation will subsequently immediately cause a partial debond.

Gamstedt and Sjögren (1999c) have identified a mechanism that accounts for the inferior fatigue behaviour in tension-compression compared with tension-tension of laminates containing transverse plies. The mechanism is debond crack propagation under compression for sufficiently large debond angles. Crack opening zones form at the crack tips under global compression. The interfacial crack can now propagate more easily than if there were contact zones at the crack tips for equal tensile far-field loading (Varna et al., 1997). The validity of this mechanism hypothesis was investigated by a number of experiments. On the macroscopic level, fatigue testing of cross-ply laminates showed enhanced transverse crack formation in tension-compression loading. On the microscopic level, model specimens with single transverse fibres were tested to examine the debond growth under low cycle tension-compression fatigue loading. It was shown that debond propagation occurred also during the compressive load excursion, due to the mismatch in elastic properties and curved crack along the interfacial debond.

3.5 Multiaxial fatigue

In a multidirectional laminate, the constitutive plies are subjected to multiaxial stresses, even though the applied far-field load is uniaxial. The neighbouring plies with other fibre orientations constrain the given ply, and due to the inherent anisotropy, multiaxial stresses readily develop although the principal stress direction (or at least the principal strain direction) is expected to be the same as the direction of the applied load. This is of course true also for cyclic loading. The inherent multiaxiality arises from the anisotropy of the constituent plies, while the external multiaxiality is brought about from the external loads. Biaxial loading in the plane of the composite is a probable scenario for large composite skin or shell structures. However, to achieve biaxial loading in test coupons to generate strength or stress-life data with true material failure without geometry effects from the specimen is not very straightforward. If out-of-plane failure can be ignored, the biaxial case is sufficient in the investigation of the effect of multiaxial loading of composite structures, which is the case for most practical applications except maybe for thick-walled composite tubes. The effect of external multiaxiality is usually greater than the inherent one.

Susmel and Quaresimin (2001) have reviewed the behaviour of composite materials subjected to multiaxial fatigue. This effect is of most interest in machine design (Chen and Matthews, 1993). The cruciform specimen geometry is a common kind to perform materials characterisation on the coupon scale.

The failure criteria are almost exclusively based on fatigue generalisation of the Tsai-Hill criterion (see e.g. Tsai, 1968). The polynominal coefficients and their cycle dependency are determined from experimental tests of varying

degrees of biaxiality. Fawaz and Ellyin (1994, 1995a) used such a method to predict fatigue life in multiaxial loading, where the stress analysis is reduced to all of the unidirectional plies that make up the laminate. If the fatigue-life behaviour of a single unidirectional ply is known for each of all possible load cases, the fatigue-life behaviour of a general lay-up would be predictable with this model. Other functional forms based on certain damage metrics have also been proposed (Kawakami et al., 1995), which was correlated to experimental data for a plain woven glass-fabric-reinforced polyester under tension and torsion biaxial loading. The common denominator for all multiaxial models found in the literature was that they were based on macroscopic stress-life data, ignoring the mechanisms and damage accumulation leading to the critical state at failure.

3.6 Post-impact fatigue

Low-energy impact in helicopter rotor blades may induce damage that could become critical after further fatigue loading, and a trustworthy non-destructive method is essential in order to detect this type of damage in time (Gros, 1996). Impact damage by birds, hailstones and debris is certainly worthwhile to consider for rotorblades for wind turbines and for aeronautical propellers. Resistance to shoal impact is most desirable for marine propellers.

Abrate (1998) treats the subject of fatigue after impact briefly in his book on impact in polymeric composite materials. It also contains a valuable list of references on which his conclusions are based. The major part of the limited literature concerns carbon fibre/epoxy or carbon fibre/PEEK and relatively thin laminates. Abrate (1998) states that although it is difficult to generalise observations made from a limited number of studies, it appears that no drastic effect is to be expected when impact damaged laminates are subjected to fatigue loading. This conclusion is to some extent supported by some of the earlier studies on the subject (Cantwell et al., 1983, 1984; Ramkumar, 1983): Both compression and tension fatigue behaviour was studied for carbon fibre/epoxy laminates with stacking sequences involving 0°, ±45° and 90° plies. The fatigue performance of the material was evaluated by its influence on the residual strength of the specimens in either tension or compression. In most cases the residual strength were lower after fatigue loading. The influence from the impact damage is generally large on the compression properties, and lower on the tensile properties. This is primarily due to out-of-plane bucking and subsequent failure in compression of laminates with impact-induced delamination.

Two different impact damages are considered by Ramkumar (1983): Type I is the BVID kind (barely visible damage), and type II is a damage involving visible cracks on the laminate surface(s). Three types of constant amplitude fatigue testing is reported ($R = 0$, -1 and $-\infty$) and compared with the corresponding results for the undamaged laminates. The reduction in fatigue threshold was 19% for tension-tension loading ($R = 0$), 30% for fully reversed tension-compression ($R = -1$), and 29% for compression-compression loading ($R = -\infty$). The noticeable larger knockdown in fatigue limit in loading modes involving compressive load excursions, can again be attributed to out-of-plane delamination and buckling. This finding is further substantiated by Portanova et al. (1992) who found a larger decrease in post-impact compressive fatigue performance of toughened tape laminates than that of through-thickness stitched uniweave laminates. The stitching has been found to suppress extended delamination on impact (Bibo and Hogg, 1998), from which further damage

growth can propagated. Delamination criticality in composite structures is particularly important in compressive loading (Olsson et al., 2000).

The impact damage development in a carbon fibre/epoxy composite were studied by Clark and van Blaricum (1987) for laminates subjected to variable amplitude loading in compressive mode. The main conclusion is that even low-level impact damage can propagate to failure by moderate compression dominated loading. The growth rate was evaluated by ultrasonic technique, which means that the damage growth observed was primarily delamination growth. The difficulties in obtaining a consistent distinction between the impact damage delamination and the fatigue growth zone is pointed out by Clark and Saunders (1991), where a de-ply technique was used to investigate the complex crack pattern of delaminations and ply-cracks in the various laminates. The delaminations seemed to grow from the ply-cracks crossover points, and the detailed fracture mode was influenced by the presence of neighbouring delaminations.

The analogy between fatigue of impact damaged specimens and specimens containing machined holes are treated by Portanova et al. (1992), Lauder et al. (1993) and Swanson et al. (1992). Two different material systems were considered by Portanova et al. (1992) for compressive loading, and reduction in fatigue life for impact damage was comparable to those reported by Ramkumar (1983). The two material systems studied were a stitched unidirectional fabric and a reference material made of toughened tape composite. For open hole specimens, the fatigue lives of the two material types were approximately equal. As mentioned previously, the fatigue lives were larger for the impact damaged specimens with stitches. A direct comparison between impact damage and machined hole in glass-epoxy laminates are reported by Lauder et al. (1993) for specimens loaded in four-point bending. It is concluded, that there is a similarity between the endurance limit of through-holes and penetration impact holes with the same area. The material volume of impact damage can be considered as a softened inclusion in the pristine laminate, where a machined hole would correspond to an infinitely compliant inclusion. Open holes may therefore be considered as an extreme case of impact damage, where the other extreme would be the intact as-fabricated laminate.

The compression fatigue response of quasi-isotropic carbon fibre laminates with two different types of matrix systems (a brittle and a toughened) is reported by Swanson et al. (1993). Both impact damaged and open hole specimens were studied. The residual compressive strength for toughened matrix system was approximately twice that of the brittle system. The fatigue loading improved the residual compressive strength for both types of matrices and specimens, which can be explained by wear-in by soothing axial delamination that decrease stress concentrations. Surprisingly little if any increase in damage area were observed as measured by C-scan technique. Other damage types may be active but not revealed by this technique. The assumption that failure in compression after fatigue is due to buckling of delaminated area makes it difficult to explain the improvement of the residual strength during fatigue loading. The buckling failure mode may be relevant under certain conditions, but it seems not to be the only one and not always the dominating damage growth mechanism leading to failure. It can be speculated that a transition from out-of-plane buckling to local kink-band formation as failure mode could explain the increase in residual strength.

Beheshty and Harris (1997, 1998a, 1998b) have discussed the extension of an empirical constant-life model to the regime of post-impact fatigue of laminates. The discussion is based on one type of laminate $[(\pm 45, 0_2)_2]_S$, and constant life diagrams is given for both carbon fibre/epoxy and glass fibre/epoxy materials. According to the diagrams, the influence on the fatigue life is very large for

compression loading and approximately similar for both types of material. In tension-tension loading, the diagrams predict a smaller influence. The underlying reason must be the impact induced delaminations that speed up subsequent damage accumulation in compressive fatigue.

The post-impact tension-tension fatigue behaviour of quasi-isotropic carbon-epoxy laminates has been studied by Tai et al. (1998). A reduction in the residual tensile strength after impact is reported, and the experimental scatter was found to become larger with larger impact energies. However, the impact damaged specimens showed a reduced fatigue life and reduced overall data scatter relative to the as-received specimens without impact damage. The material volume damaged by impact serves as initiation 'point' and confines further damage development in a smaller volume than in the as-received material. The reduction in fatigue life is significantly larger at lower stress levels than at higher levels. A reduction in fatigue load of 8-10 % at a lifetime of 10^5 cycles can be deduced from the data given by Tai et al. (1998). Low-energy impact damage showed a significant effect in high-cycle (low load) fatigue testing. This underscores the role of impact damage as an instigator of damage accumulation leading to fatigue failure.

In summary of post-impact fatigue, it can be mentioned that relatively few papers are published on the subject. All the studies were performed on specimens and materials, which were unstressed during impact. The loss of strength under simultaneous impact and tensile stress can be much more severe than under the sequential procedure of impact followed by tensile stress, which is indicated by Philips et al. (1990). This effect may be important for structures subjected to constantly high stresses such as rotor blades for wind turbines and helicopters, aircraft and watercraft propellers, etc.

4 Predictive methods

4.1 Damage accumulation

As can be seen from this review, there are ample reasons to develop suitable test methods to characterise the fatigue behaviour of composites used in rotating structures for improved predictions of property degradation and lifetime. The understanding of failure modes and underlying damage mechanisms needs to be improved considerably, in order to achieve more optimised design and selection of materials. Mechanism-based models should be striven for, since they convey a relation between the structure and performance of the material. This gives possibilities to some degree of extrapolative predictions and eventually materials design or optimisation.

The main types of models are based on (*i*) changes in stiffness, (*ii*) the degradation of the residual strength, and finally and preferably (*iii*) the actual damage mechanisms. In this section, the latter case is considered. Due to the complexity of fatigue damage mechanisms, mechanistic models that account for damage accumulation all the way to failure are rare and hardly of general use. The scientific efforts have instead been focused on clarifying the process of damage accumulation for some simple cases. The most frequent subject of investigation is the accumulation of transverse cracks in tensile fatigue of cross-ply laminates. The motivation is clearly that the simple cases must first be fully understood before one could move on to more general lay-ups and types of loading in which a multitude of interacting damage mechanisms are active.

In homogeneous materials with propagating cracks, the Paris law is frequently used. It may also be used to account for the kinetics of various types of damage, such as delaminations and transverse cracks, in composites. Paris law, see Equation 1, is a phenomenological relation described by a power law whose parameters are fitted to experimental data. Nevertheless, empiricism in the treatment of the growth of microscopic damage provides a more mechanistic description than empiricism of macroscopic behaviour, since fewer kinds of damage phenomena have to be characterised. Furthermore, this means a possibility to generalise to a wider range of macroscopic configurations.

Transverse cracking in 90° plies and interlaminar delaminations starting from transverse cracks along the interfaces to adjacent plies have been observed in static and fatigue experiments (e.g. Takeda and Ogihara, 1994). In contrast to wear-out models of macroscopic scalar damage measures, as will be discussed in the subsequent section, the damage development in the composite can now described by the interaction of two types of damage, namely transverse cracking (TC) and delamination (DL). The macroscopic scalar restriction of a vectorial representation of damage state will certainly impair the accuracy of any model. A model that accounts for the accumulation of these two types of observed damage is certainly more realistic, and should be more precise as well as reduce the degree of empiricism. By stress analysis the energy release rates for each of these two mechanisms, G_{TC} and G_{DL}, can be determined (Nairn and Hu, 1992). As shown by Takeda et al. (1995), the kinetics of both kinds of damage can be expressed independently by a Paris law, although they are mutually dependent on the stationary extent of the other kind of damage,

$$\frac{\mathrm{d}\rho}{\mathrm{d}N} = C_{TC}\left\{\Delta G_{TC}(\rho,\bar{l})\right\}^{m_{TC}}, \tag{2}$$

for transverse cracking, and

$$\frac{\mathrm{d}\bar{l}}{\mathrm{d}N} = C_{DL}\left\{\Delta G_{DL}(\bar{l},\rho)\right\}^{m_{DL}}, \tag{3}$$

for delamination, where ρ and $\bar{l}$ are the transverse crack density and average delamination length, respectively. The C and m parameters are determined by fitting Equations 2 and 3 to experimental data from microscopical observation of transverse cracking and delamination processes. The progressive behavior of these mechanisms is reflected by the m parameter. It has been found the $m_{TC} \gg m_{DL}$ by several orders of magnitude (Takeda et al., 1995), which indicates that transverse cracking is a relatively initiatory mechanism and delamination is a predominantly progressive. The propensity of delamination in the presence of transverse cracking has been illustrated by Selvarathinam and Weitsman (1998) in terms of energy release rates. Once the parameters have been characterised, the evolution of damage can be predicted for any load history. From the damage measures, the stress history of the critical element of the laminate, i.e. the one that is expected to fail lastly, can be obtained. It would then be possible to predict global lifetime if fatigue-life data for the critical element, typically the 0° ply, are provided (Akshantala and Talreja, 2000).

Other types of mechanisms are likely to influence the accumulation of damage in addition to transverse cracking and interlaminar delamination. Longitudinal splits in the on-axis plies and local delaminations growing radially from intersection of longitudinal splits and transverse cracks can be expected under more realistic conditions (Henaff-Gardin and Lafaire-Frenot, 2001). Matrix shear yielding and cracking in laminates containing [±45] elements are

another possibility that must be duly taken into account (Tong et al., 1997). In finite width specimens, delaminations originating from the specimen edges are expected and should also be taken into consideration (Ogihara et al., 1999). This feature makes the damage accumulation highly dependent on specimen geometry. In other words, the accumulation of damage is not exclusively controlled by the material and stress level. However, the underlying mechanisms and their controlling rates provide a material description, and could be used to predict damage development in a more general sense.

This illustrates the complexity of fatigue damage process in composite laminates, and highlights that much work remains to be done - both in terms of modelling and, especially, supporting microscopic experimental work. The reward would be an immense improvement in reliable fatigue design compared with phenomenological approaches.

4.2 Fatigue life prediction

Most life prediction models are empirical and are based on macroscopic measurements, to which the parameters of the model are fitted. A review of frequently used models has been compiled by Kienert (1999), Davies and Curtis (1999), and by Nyman (1996). There are many relations for predicting fatigue life such as power laws like Baquin's relation, straight fits to S-$\log N$ data, Coffin and Manson's relation. They are all empirical fits to stress-life data. Tolerance and confidence bounds are extracted from data reduction of the S-N data (Ronold and Echtermeyer, 1996; Echtermeyer et al., 1996). The next more advanced description, which is frequently used in practice, is models of residual strength degradation. It is generally based on the static strength distribution and a deterministic empirical equation for the degradation of this strength. The failure condition is given when the residual strength has decreased to the value of the applied stress. Such degradation models have been proposed by e.g. Whitney (1981), Rotem (1988), Hwang and Han (1989), and Schaff and Davidson (1997a, 1997b).

Originally, the S-N relation was naïvely calculated from linear regression of the fatigue life data for constant load amplitudes. A linear regression scheme requires a normal distribution of the fatigue life. However, the normal distribution is a mediocre representation of the scatter in fatigue life. The Weibull distribution is generally accepted as the best representative to describe fatigue life of composites (Hwang and Han, 1989). Procedures to generate linear S-N curves from fatigue lives characterised by a two-parameter Weibull distribution have been put forth by Whitney (1981).

More refined models with non-linear S-N curves, are usually obtained by linking the fatigue life to an observable damage metric, D. The basic principles of the empirical degradation scheme will be outlined in the following. The evolution of the damage metric D is monotonic during fatigue. Typically, D represents axial residual strength (Sendeckyj, 1981), axial Young's modulus (Hwang and Han, 1989) or damage size (Beaumont, 1987). This degradation can be represented by a differential equation:

$$\frac{\mathrm{d}D}{\mathrm{d}N} = f(\sigma, R, D), \tag{4}$$

where f is a function the applied maximum stress, σ, stress ratio, R, and the damage metric, D. Kachanov (1986) introduced the damage variable D (originally ω) as a scalar. A tensorial representation with a more detailed description of the damage state can be found in work by Talreja (1994).

However, for many practical applications in engineering usage, a scalar damage variable might be a more suitable choice, since the damage state relates to one of the aforementioned physical quantities which is macroscopically measurable.

The residual strength degradation model is the most frequently employed one in the literature. The degradation expressed in the previous equation is rewritten in terms of a monotonically decreasing residual strength, i.e. a wear-out model. The degradation rate is typically assumed to obey a power law of the form

$$\frac{\mathrm{d}\sigma_r}{\mathrm{d}N} = -\frac{A\sigma_r}{\alpha\sigma_r{}^{\alpha-1}}, \tag{5}$$

where σ_r is the residual strength, and A and α are fitting parameters. The degradation function on the left-hand side of the equation is chosen ad hoc. The applicability of a number of functions of similar kind has been tested by Hashin (1985), and Hwang and Han (1989). The specimen fails when the residual strength decreases to the applied maximum stress , i.e. $N = N_f$ when $\sigma_r = \sigma$. The value of N_f is obtained by integration of Equation 5. This deterministic criterion can also be formulated in a statistical framework to account for the scatter in lifetime data (Talreja, 1990).

Carbon-fibre reinforced plastic laminates with a large proportion of unidirectional 0° plies are known to have a more abrupt decrease of the residual strength at the point of failure than the smoothly decreasing strength given by the above degradation equation (Chou and Croman, 1979; Adam et al., 1986; Ma et al., 1995). The 'sudden death' model formulates an abrupt decrease of the residual strength at failure, which means that the residual stiffness remains constant until the point of failure.

For competing failure modes, e.g. tensile and compressive failure in tension-compression cycling, a system of differential equations of the type in Equation 5 must be considered (Whitney, 1983). Each equation corresponds to a specific failure mode. Indications of competing failure modes generating clusters in the S-N diagram have also been observed for unidirectional composites in tension-tension loading Barnard et al. (1985).

The stress amplitude of a composite member may vary due to stress concentration, stiffness change in neighbouring plies because of damage accumulation, or application of block or non-stationary spectrum loading. The constant amplitude models should in this case be modified to account for variable amplitudes. When the stress changes from one level to another, the damage state remains momentarily constant. The damage state uniquely defines position in the fatigue life, and damage accumulates progressively during fatigue. The variable amplitude models can also be termed cumulative damage models.

The classic example of this approach is the Palmgren-Miner rule (Palmgren, 1924; Miner, 1945), in which the damage metric is set to the fractional life at the different cyclic stress levels, and failure is assumed to occur when the damage sum exceeds unity, i.e.

$$D = \sum_i d_i = \sum_i \frac{N_i}{N_{fi}}, \tag{6}$$

where N_i and N_{fi} are the number of cycles performed and cycles to failure at stress level i. The Palmgren-Miner failure criterion is irrespective of the order of application of stress level blocks, and is known to yield non-conservative predictions for composites (Curtis, 1989; Adam et al., 1994). Testing in compression-compression fatigue followed by tension-tension fatigue is more

detrimental than the opposite order (Adam et al., 1994). This calls for a non-linear cumulative damage rule which accounts for load block order. The recursive Marco-Starkey model is more suitable for composite materials in which the damage mechanisms are highly dependent on the applied stress level (Marco and Starkey, 1954). Furthermore, one type of damage may incite enhanced activity of another damage mechanism, and the order of block loading is hence of primary concern. The Marco-Starkey model, which is based on the iso-damage concept and is a generalization of the linear Palmgren-Miner model, can be expressed as

$$d_i = \left(\frac{N_i}{N_{fi}} + d_{i-1}^{\gamma_{i-1}/\gamma_i} \right)^{\gamma_i}, \tag{7}$$

where γ_i is a function of stress amplitude, and $d_0 = 0$ reflects the pristine state. Failure occurs when $d_i = 1$. Similar non-linear cumulative damage models have been proposed by Hashin (1985), Reifsnider and Stinchcomb (1986), and Gao (1994). The model proposed by Reifsnider and Stinchcomb has evolved to the commercial software package MRLife for fatigue life predictions. The mechanistic background of the sequence effect in variable amplitude loading has been addressed by Gamstedt and Sjögren (2001b).

For continuous variation in the stress amplitude level, Equations 6 and 7 may be adapted to a differential form. Non-linear damage accumulation rules have successfully been used to predict the fatigue life of multidirectional laminates containing at least one 0° ply (Reifsnider and Stinchcomb, 1986; Diao et al., 1995). The global stress level has here a constant amplitude. As subcritical cracks in the off-axis plies develop, the local stress state in the 0° ply changes continuously. When the 0° ply fails, the entire laminate fails. The 0° ply constitutes the critical element of the laminate. The critical element models are based on the cumulative damage models since the local stress amplitude varies. Experiments and satisfying predictions of the critical element model underscores the relative importance of the 0° ply (Gao, 1994). A critical element model considering localised damage in the 0° plies close to the transverse cracks in the adjacent plies would be a more physical description. However, delaminations and longitudinal splitting in the region of the critical model makes it difficult to predict the evolution of the local stress amplitude as was seen in the previous section.

One of the problems with the empirical variable amplitude models is that the chosen damage metric, usually a scalar determined from macroscopic testing, does not reflect the true damage state of the composite. A set of damage parameters that relate in more detail to the current damage state would be more adequate (Talreja, 1994). This can only be done by identifying the dominating fatigue damage mechanisms, and characterising their growth rates and critical values as a function of load level.

The equal rank of strength and life is an important notion and plays a significant role in fatigue modelling of composites. It is postulated that if a sample of components could hypothetically be tested for both static strength and fatigue life, each individual member would occupy the same rank in both strength and fatigue life data sets. The percentiles in strength, σ_p, and life, N_{fp}, are thus directly linked by the fatigue life relationship;

$$\frac{\sigma_p}{\sigma} = F(N_{fp}), \tag{8}$$

where σ is the applied stress, and F is a monotonically increasing function dependent on the fatigue life. By fatigue life measurements and estimation of the distributions of strength and life, F can determined (Barnard et al., 1988). Similarly, the life distribution can be deduced from the fatigue life curve and the strength distribution (Barnard et al., 1985). With this above scheme, the deterministic cumulative damage models can be modified for the analysis of life distributions (Hwang and Han, 1987). Wang et al. (1981) showed that proof tests can ascertain a minimum life with a reasonable confidence. This supports the assumption of equal rank. However, proof tests can not be recommended since they are likely to inflict some damage from which fatigue damage can propagate, and thereby reduce the fatigue resistance of the material. A non-destructive technique for quality control is a better option.

An argument against the equal rank assumption is that the damage mechanisms differ in fatigue at lower stress levels and in static testing. It can therefore not be excluded that the fatigue degradation and subsequent failure may occur earlier for specimens with lower static strengths. However, for stress levels close to the static failure stress where similarities in failure mechanisms are notable, the assumption could be considered valid. The strength life equal rank assumption seems to be valid in practice for most materials, but its general truth is questionable.

The predictive methods discussed above are not really based on mechanisms. If the more detailed methods presented in the previous section on damage accumulation could be complemented by a failure criterion, a more accurate method for life prediction could be conceived. This would increase the reliability in life prediction. Research is urgently required to develop such models and design tools that can be made widely available to composite designers. In contrast to phenomenological models, the mechanistic approach would require less experimentally obtained input. The experiments would be on a small scale, tracking the growth of damage. The primary drawback of the phenomenological approaches with a strong dependence on large amounts of macroscopic experimental input for each material, lay-up and loading conditions could then be avoided.

4.3 Scaling effects

Scaling of strength and fatigue behaviour from test coupons to large structure is of great importance in design. In practice, the fatigue behaviour measured on small laboratory specimens is used directly as input in much larger structures together with a large, but arbitrary selected, safety factor. As for any brittle material, the strength of composites is controlled by the largest flaw or defect. Since the probability of finding a larger flaw, meaning lower strength, is higher for a larger volume, the strength will scale together with volume. This was first recognised and formulated by Weibull (1939).

The use of Weibull theory to predict volume-dependent strength requires knowledge of the spatial stress distribution in the volume V of the material. A statistically significant population of strength samples for the volume is also necessary. If the stress distribution of another volume is known, the strength can be predicted. The probability that the material can survive a stress distribution $\sigma(x, y, z)$ is given by

$$P_{\mathrm{S}} = \exp\left\{-\int_V \left[\frac{\sigma(x,y,z) - \sigma_{\mathrm{u}}}{\sigma_0}\right]^m \mathrm{d}x\,\mathrm{d}y\,\mathrm{d}z\right\}, \tag{9}$$

where σ_u is the threshold stress below which the material would not fail, m is the Weibull modulus or shape parameter, and σ_0 is the scale parameter. For varying volumes V, the strength distribution can now be scaled with the other material parameters kept constant.

In static strength, Jackson et al. (1992) noticed a large scale effect in both strength and strain to failure for composites, whereas Lavoie et al. (2000) did not find any size scale effect. Their results indicated that apparent scale effects may be attributed to failure close to the end grips due to the stress concentration from gripping. Care must thus be taken only to use data where failure has actually occurred within the gauge length. It was also shown that straightforward use of the scale parameter m estimated from strength values for one size specimens can not be applied directly in the scaling to other sizes. Apparently, the scaling data for different specimen sizes yields smaller values of m, indicative of larger scatter in strength. In lack of physical models correlated to microscopic experimental characterisation, the scaling effect has to be accounted for by a relatively large amount of macroscopic tests. However, some theoretical work of the size scaling of strength has been done (Curtin, 1998), in which simulations of the details of fibre breakage are performed. The results are in accordance with the macroscopic approach, but provide better understanding of the controlling microscopic parameters. Similar Monte Carlo simulations by Mahesh et al. (1999) show that the stress-redistribution mechanisms around fibre breaks influence the size scaling effect significantly. Microscopic experiments to support simulation studies would be most useful in the quest for applicable mechanism-based models.

There is a striking lack of work on scaling effects in fatigue. However, Chawan and O'Brien (2000) have noticed a volume effect in bending fatigue of a glass-fibre-reinforced epoxy which is explained by the classical weakest link effect. It is not clear how the scaling effect in fatigue would compare the scaling in static strength. It can be speculated that the scaling effect is less severe in fatigue loading since fatigue failure tends to be more 'ductile' with larger damage zones, whereas static failure is more 'brittle' and controlled by the size of the largest inherent flaw of the material. If fatigue reliability methods should be improved for design of composite structures, the scaling effect in fatigue must be investigated carefully. Large safety factors and over-conservative dimensions could then be reduced, which also means a considerable reduction in cost.
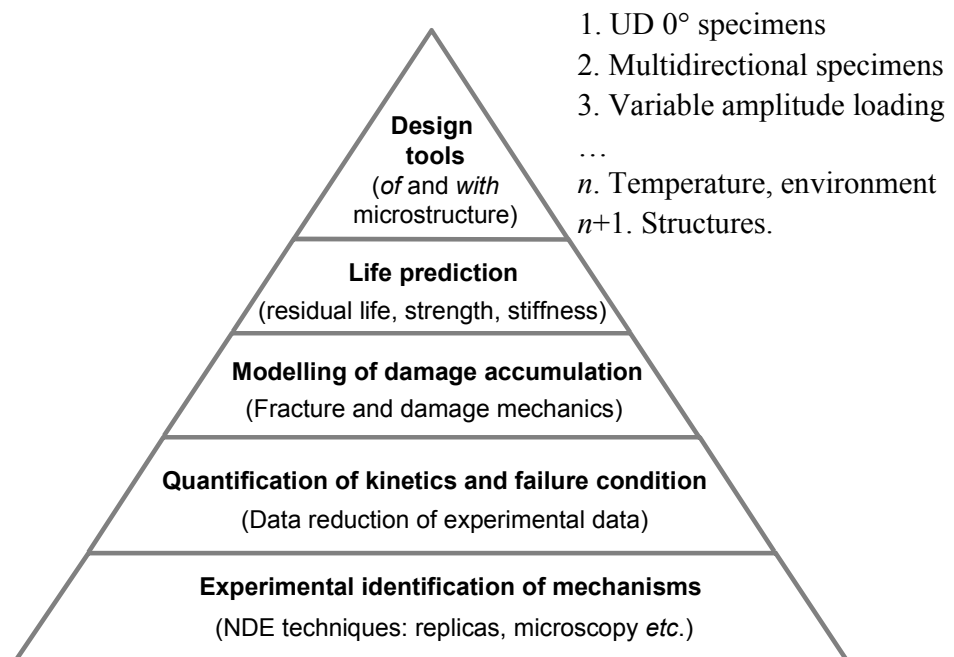
4.4 Mechanism-based modelling

The lion's share of fatigue life prediction schemes presented in the technical literature concerns degradation of a macroscopic property, i.e. residual stiffness, residual strength or residual life, described by an empirical rate until final failure occurs. A fatigue damage model based on the microscopic mechanisms would require less costly testing in the long run, and provide a better understanding, which would give indications of improvement of material properties. The identifications of mechanisms presented in this manuscript are but a tiny step toward the ultimate goal of fatigue life prediction. It should be emphasised that a mechanism-based fatigue model demands extensive and precise experimental work to identify and quantify the operative fatigue damage mechanisms, from which kinetic relationships intrinsic to the composition of a given composite material could be deduced. For life prediction, the kinetic equations must be complemented with a failure criterion. A design tool against fatigue could thus be conceived. This link is still qualitative. Simple material structures like unidirectional longitudinal specimen in tensile fatigue should

first be analysed quantitatively, before generalising on to more complex scenarios.

It is useful to keep the goal of the top-down approach in mind, but practically a bottom-up approach is more logical. Once the micromechanisms have been understood, as indicated in the fundament of the pyramid in Figure 15, we can move on upwards in the structural hierarchy towards damage mechanics on the ply-level dealing with features like transverse cracking and delamination, see e.g. Akshantala and Talreja (1998, 2000). It would then be feasible to conceive life prediction schemes which can be used for engineering design, while maintaining the link downwards to the microstructural mechanisms. For mechanistic life prediction of unidirectional composites, a similar approach as the one taken by Phoenix et al. (1988) and Otani et al. (1991) can be advocated. They were concerned with creep in composites with few fibres, but the time-dependent matrix creep between the fibres is not unlike the fatigue scenario portrayed in Figure 7 with progressive debonding. Ibnabdeljalil and Curtin (1997) has developed Green lattice models that account for matrix shear failure and stochastic fibre rupture in unidirectional composites, which could be extended to fatigue.

Active research in the composite community is focused to establish the discussed link from mechanisms to properties. Qualitative and illustrating examples of fatigue damage mechanisms and their effects have been presented in this manuscript. For the reader who likes to become more acquainted in quantitative prediction of damage accumulation, the articles cited in the reference list can be consulted.

1. UD 0° specimens
2. Multidirectional specimens
3. Variable amplitude loading
…
n. Temperature, environment
$n+1$. Structures.

Figure 15. Bottom-up approach to fatigue in composite materials, starting from the damage mechanisms that are controlled by the microstructure and processing conditions, continuing by sensible modelling to the macroscopic engineering properties.

5 Concluding remarks

The present report covers a wide span of disciplines, and is therefore far from comprehensive. The goal has been to review some of the most important issues concerning fatigue degradation and failure of rotating structures. Particular emphasis has been placed on test methods for materials characterisation, and how the damage mechanisms give rise to the measured properties. Some suggested guidelines of suitable directions for further research in this industrially important area are outlined. The potential benefits from a better understanding of fatigue degradation of these applications are paramount if more accurate design tools could be developed based on physical models that take the structure of the material into account.

Since the fatigue scenario is similar for many rotating composite structures, a joint approach to materials characterisation and understanding of damage mechanisms can be undertaken. The missing link in mechanism-based models to describe property degradation and predict lifetime is experimental data on the accumulation of damage. This is the necessary input in conceiving useful models. As a variety of non-destructive characterisation techniques have matured in the last couple of years, they provide a very useful and essential asset in the experimental work. Thereafter, predictive models can be formulated and validated or rejected by the experimental data. In the years to come, we will see a shift from phenomenological fatigue models for composite materials towards mechanistic models that require less costly but more precise and detailed experimental input. This will yield more accurate predictions, and thereby warrant more widespread adoption of polymer composites in primary structures, by significantly reducing over-design and large factors of safety.

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Title and authors

Fatigue degradation and failure of rotating composite structures –
Materials characterisation and underlying mechanisms

E. Kristofer Gamstedt and Svend Ib Andersen

Abstract (max. 2000 characters)

The present review concerns rotating composite structures, in which fatigue
degradation is of key concern for in-service failure. Such applications are for
instance rotor blades in wind turbines, helicopter rotor blades, flywheels for
energy storage, marine and aeronautical propellers, and rolls for paper
machines. The purpose is to identify areas where impending efforts should be
made to make better use of composite materials in these applications. In order to
obtain better design methodologies, which would allow more reliable and
slender structures, improved test methods are necessary. Furthermore, the
relation between structural, component and specimen test results should be
better understood than what is presently the case. Improved predictive methods
rely on a better understanding of the underlying damage mechanisms. With
mechanism-based models, the component substructure or even the material
microstructure could be optimised for best possible fatigue resistance. These
issues are addressed in the present report, with special emphasis on test
methods, and scaling from damage mechanisms to relevant material properties.

Descriptors INIS/EDB

COMPOSITE   MATERIALS;   DAMAGE;   FAILURES;   FATIGUE;
MATERIALS TESTING; REVIEWS; ROTORS; TURBINE BLADES