Handbook of Properties of Textile and Technical Fibres

Edited by A. R. Bunsell
Handbook of Properties of Textile and Technical Fibres
The Textile Institute Book Series

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Edited by

Anthony R. Bunsell

The Textile Institute

Elsevier
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Tensile fatigue of thermoplastic fibers

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16.1 Introduction

Fine-diameter thermoplastic fibers, principally polyamide (PA) and polyester fibers, are used for traditional textile applications as well as in advanced technical structures. The failure of the fibers due to the repeated loading of the fabric or structure can, in some cases, have serious consequences and the mechanisms leading to unexpected fatigue failures have to be understood and taken into account for many applications.

The fibers that will be discussed in this chapter will be melt-spun thermoplastic polyester (polyethylene terephthalate [PET] and polyethylene naphthalate [PEN]) and nylon (PA 6 and 6.6) fibers. PET fibers are the most widely used and produced fibers throughout the world. PA fibers were the first synthetic fibers to be produced. Both types of fiber find wide use in apparel and in high performance technical structures such as tires, cables such as mooring ropes, parachute cords, and belt drives. There are many other examples of course. The desirability of avoiding unexpected failure, due to fatigue, in the latter structures should be obvious. These types of fibers fail by fatigue under certain types of cyclic tensile loading. The distinctive fracture morphologies, which occur when fibers fail in fatigue, can be used for diagnostic purposes and allow an insight into the mechanisms controlling this behavior. The understanding of the fatigue processes in these fibers can suggest ways of reducing or eliminating the probability of unforeseen failures.

A particular interest is the use of reinforcing thermoplastic fibers in tires. The tire is the oldest composite developed with an elastomeric matrix and multiple layers of reinforcement. An example of the structure of a tyre is shown in Fig. 16.1.

Initially cotton plies were used as reinforcement but were replaced by regenerated cellulosic yarn and then by both PET and PA 6.6 fibers.

All these modifications were driven by the very challenging thermomechanical conditions to which the cords were subjected. Depending on their position inside the tire, the cords are subjected to a combination of strain-imposed compression as well as stress- or strain-imposed extension and bending. All this occurs under various thermal conditions with typical orders of magnitude from −40°C (aircraft tires) to 100°C in current usage in the carcass plies for tires and more than 120°C in the outer layers, particularly for high performance tires. As the cords are made from twisted yarns,
this leads to local pressure in the cords, which could also be an issue adding a transversal mechanical component.

If for example, a passenger car tire travels for a distance of 40,000 km, the product experiences approximately 20 million cycles with very varying conditions of load, air internal pressure, and environmental temperature. These cycles could locally induce extension, compression, and bending of filaments. A good understanding of the tire mechanics requires a detailed description of the main phenomenon so as to obtain the best performance from each component. It must include the understanding of the influence of intrinsic parameters (polymer nature, microstructure, additives) and extrinsic parameters (load, cycles, temperature) on tensile fatigue of reinforcing fibers.

PET and PA fibers are drawn from the melt, and the act of drawing aligns the fiber molecular structure, enhancing the fibers’ properties and making them anisotropic. The molecular structure in these organic fibers is described in detail in Chapters 12 and 13. An undrawn organic thermoplastic fiber would have low rigidity and strength but high elongation. It would also show exaggerated plastic deformation, which would not allow it to be used for most applications. The alignment of the molecular structure, during drawing, means that more of any load applied to the fiber is supported by molecules that are parallel to the fiber axis. This means that their deformation is controlled by the rigid first-order covalent atomic bonds in the molecular skeleton backbone rather than more compliant secondary bonds, such as hydrogen and van der Waals forces, which determine intermolecular linkages. This process of making the fiber, from the molten polymer, does not lead to a perfect alignment of the molecular structure. The molecules are not all arranged perfectly parallel to the fiber axis. If this were the case, the rigidity of such fibers would be much greater. For this reason other manufacturing processes have been used to produce fibers with their macromolecules aligned much more parallel to the fiber axis and the results are remarkable.
increases in mechanical and often heat resistance properties. This type of fiber is used for technical apparel and high-performance fiber—reinforced composites including high-performance tires. These fibers are usually made by a liquid crystal process in which molecular alignment occurs intrinsically due to atomic bonding within, usually, a solution and the locally aligned molecules are then arranged naturally parallel to the fiber axis during passage through a spinneret, as discussed in Chapter 17. Fatigue can be an issue with these fibers but has been less studied than for thermoplastic fibers treated in this chapter. One reason is the high anisotropy of liquid crystal spun fibers, which leads to extremely complex fibrillated failures, which are difficult to interpret.

16.2 Principles of tensile fatigue

Fibers are long fine structures. Conventional thermoplastic fibers have diameters usually in the range from 5 to 40 μm and the technical fibers that will be discussed here have diameters around 10–25 μm. This can be compared to the diameter of a human hair, which is about 80 μm. Their fineness means that even the stiffest fibers in tension are very flexible in bending, as demonstrated in Chapter 1. The bending stiffness is a function of the reciprocal of the fiber diameter to the fourth power. Although there are ways of getting around the buckling of fibers in compression, most evaluation of fibers is in tension (Bunsell and Schwartz, 2000). The tensile fatigue evaluation of fibers presents particular difficulties as the tests are not quick to undertake and the fiber properties change throughout the test.

The study of the fatigue behavior of materials, in general, began in earnest in the 1950s, due to major problems encountered by the first civilian jet airliners. Nevertheless the failure of metallic structures under cyclic loading had been encountered since the early days of the industrial revolution. Metals however have an elastic region in which a cyclic strain of the material induces a cyclic stress. The two are in phase as the material is in its elastic domain and this means that the standard way of fatigue testing metals is to apply a cyclic deformation, which induces an in-phase cyclic load, related by Hooke’s Law, and neither is varied throughout the test. It can be noted that in most applications it is a cyclic load to which a material is subjected rather than a cyclic strain.

Early attempts to evaluate the fatigue characteristics of organic fibers initially used the same type of tests as was being used to test metals. That is to say, a cyclic deformation was applied to the fiber. As the fibers were not purely elastic, the plastic deformation produced on each cycle led to an ever increasing length, resulting in a reduced load amplitude experienced by the fiber and eventually its complete buckling. This type of simple extension cycling is shown schematically in Fig. 16.2(a).

In this test the fiber either fails in the first cycle or not at all as the maximum loading levels quickly fall.

A more complicated version of this test, designed to avoid the accumulation of plastic deformation is accumulated extension cycling and the results are shown in Fig. 16.2(b). In this test the fiber is held vertically and the plastic deformation produced on each strain cycle is removed by opening the bottom grip. A small weight attached to
the bottom end of the fiber, which passes through the lower grip, pulls the fiber taut. The bottom grip first opens and then closes and the fiber is taken through another strain cycle. This means that the volume of fiber being tested is progressively decreased so that although the maximum displacement imposed does not change, the fiber is progressively taken up its stress-strain curve. With this second type of test, the fiber ultimately fails but it can never be clear if the break is due to a fatigue process or just because the end of the stress-strain curve has been reached. The optimal way of conducting a fatigue test on an inelastic fiber is to monitor the maximum cyclic load and maintain it constant (Hearle, 1967). This requires a machine capable of adapting the loading conditions on the fiber as it creeps and deforms plastically as shown in Fig. 16.1(c) (Bunsell et al., 1971). The fiber does continue to deform by creep but this can be evaluated by constant load tests. Failure by creep under cyclic conditions, during which the fiber is subjected to the maximum load for only a brief part of the cycle, would be expected to occur after longer times than that observed if the maximum cyclic load was applied constantly. It is this maximum load cycling technique that has revealed the fatigue process in organic fibers (Bunsell and Hearle, 1972). This type of testing is usually conducted with a horizontal testing machine for ease of manipulation of the fiber specimens and that also allows tests in a liquid held in a bath and also tests at temperature with the addition of a heating coil.

16.3 The tensile and fatigue failures of thermoplastic textile fibers produced by melt spinning

16.3.1 Tensile and fatigue fracture morphologies

Fig. 16.3 shows a PA 6.6 fiber undergoing tensile failure. In this fiber, two cracks (1 and 2 in Fig. 16.3(a)) have initiated at the surface. Crack 2 can be seen to have developed across the diameter of the fiber (Fig. 16.3(b)). As the crack propagation is slow,
or stable, the plastic deformation ahead of it results in its opening (Fig. 16.3(c)). When the remaining load bearing cross-section can no longer support the load (Fig. 16.3(d)), the fiber fails resulting in two similar complementary fracture surfaces. Fig. 16.3(e) shows those complementary ends of a PA 6.6 fiber broken in tension. The arrows show the region of crack initiation.

The type of failure shown in Fig. 16.3(e) is characteristic of room temperature tensile and creep failure of PA6, PA 6.6, PET, and PEN fibers. There are two obvious regions of crack propagation (Hearle et al., 2000). From the region of crack initiation there is a bevelled zone, resulting from a phase of slow crack growth during which the plastic deformation ahead of the crack leads to an opening of the crack. The load bearing cross-section of the fiber is reduced by the advance of the crack and finally fails in an uncontrolled manner resulting in a fracture zone normal to the fiber axis direction. In the image shown in Fig. 16.3(e) the fracture morphology suggests that the fiber underwent local melting due to energy dissipated within the insulated material.

In fatigue, the difference in fracture morphology from that obtained in tension or creep is very clear. Crack initiation, as in tensile failure, is usually in the region of the fiber surface. It will be demonstrated below that initiation occurs just underneath the surface (Fig. 16.4(a) and (b)), but instead of progressing across the fiber, the crack
begins to run along the fiber at a slight angle to the axial direction (Fig. 16.4(c) and 4(d)) (Herrera-Ramirez, 2004). Arrows in Fig. 16.4(c) and 4(d) indicate the path followed by the crack during its propagation. Fig. 16.4(e) is a top view of Fig. 16.4(d).

Fig. 16.5(a) shows a schematic view of a fiber (axial cut) representing the crack propagation and the final failure. Fig. 16.5(b) outlines the reduction of the load bearing cross-section.

The final stage seen in the fatigue fracture morphology of a PA 6.6 fiber is similar to that shown in Fig. 16.3(e) in a simple tensile test. Fatigue tests are normally conducted at 50 Hz at 21°C, although, as will be explained below, raising the temperature has a marked effect on the kinetics of crack initiation and changes the fracture morphology that is obtained. The choice of 50 Hz as a testing frequency was made as many technical structures such as tyres and parachute cords experience cyclic loadings at such a frequency.

It can be seen from Fig. 16.5 that the break leaves a concave impression on the end from which the tongue of material is removed. This is different from the convex surface seen in the case of peeling of the fiber. When the load bearing cross-section is sufficiently reduced (Fig. 16.5(b)), the PA fiber fails from the root of the fatigue crack by a
tensile mechanism, as can be seen in Figs. 16.6 and 16.7, in which the two regions of tensile failure can be observed at the point of final failure.

Exactly similar tensile and fatigue behaviors are seen with PA 6 fibers with indistinguishable fracture morphologies seen in both types of nylon fiber. This is a remarkable feature of the tensile fatigue of these thermoplastic fibers that the angle of fatigue crack penetration seems to be common amongst the PA fibers and independent of draw ratio. The angle is different, as we shall see, from that observed with the polyesters, both PET and PEN, which themselves share a common type of fatigue crack growth. The PA and the polyester fibers share many characteristics in their fatigue behavior but the striking differences between the angles of fatigue crack penetration into the fiber must reflect some intrinsic difference, perhaps at the level of the molecular structure or morphology. This has yet to be fully understood.
Analogous fatigue behavior, at room temperature, to that observed with PA fibers is seen with polyester (PET and PEN) fibers both in tension and fatigue; although, as mentioned above, some differences in the angle of crack penetration into the fiber and also final failure, occur in the latter cases. The failures of PET fibers in tension or creep give very similar fracture morphologies to that shown for PA fibers in Fig. 16.3(e). In room temperature fatigue, the same scenario of initiation near the surface followed by propagation along the fiber, gradually reducing the load bearing cross-section is again observed. However, the angle of propagation is smaller than in the PA fibers, leading to a longer crack before the load bearing section is sufficiently reduced to cause failure, as can be seen in Fig. 16.8, which shows a highly drawn technical PET fiber of 18 μm in diameter, which has failed in tensile fatigue at room temperature after cycling from zero load to 75% of simple tensile strength during $1.51 \times 10^6$ cycles.

The final failure stage of a PET fiber, which breaks in fatigue, occurs behind the fatigue crack tip by a creep process. The initiation of the final creep failure phase does not occur from the fatigue fracture surface but from near the apparently undamaged fiber surface, as shown in Figs. 16.9 and 16.10.

The failures of polyethylene naphthalate (PEN) fibers in tension and fatigue seem identical to those of PET fibers and the same long breaks are seen under cyclic loading, which leads to fatigue (Lechat et al., 2006). However, the PEN fibers do show a mechanism that may exist in PA and PET fibers but is less easily observed in these latter fibers. Under certain cyclic loading conditions PEN fibers fail with an apparent creep fracture morphology; however, a closer inspection of the slow crack growth zones

**Figure 16.7** Fatigue failure occurs in PA 6.6 fibers when the load bearing cross-section is sufficiently reduced to produce tensile failure, showing two regions of crack growth of slow and rapid propagation.
Figure 16.8 The stages of tensile fatigue failure in PET fibers are similar to those seen in PA 6.6 fibers but the angle of crack penetration into the fiber is much smaller, which leads to much longer fractures.

Figure 16.9 Final stage of room temperature fatigue failure of a PET fiber revealing that failure occurs behind the fatigue crack tip and is produced by creep of the load bearing cross-section.
reveals that a step by step growth of the crack has occurred, shown as a series of striations (Lechat et al., 2006). The striations can be seen in Fig. 16.11.

Figure 16.10 The final stage of failure in a fatigued PET is initiated at or near the surface and not at the fatigue-induced crack. This final process is governed by creep failure of the reduced load bearing cross-section of the fiber.

Close inspection of fracture morphologies of PET fibers, originally interpreted as being due to creep under cyclic loading conditions has revealed that faint striations, similar to those seen in PEN fibers can sometimes be observed. It is not known why

Figure 16.11 Striations seen in the apparent creep or tensile failure morphology of PEN fibers tested at 50 Hz.
the striations are so much more obvious in PEN fibers but their identification suggests the possibility of another type of fatigue crack growth in fibers. Another explanation would be that the striations are due to arrested slow tensile cracks that stop as the load falls, due to the cyclic form of loading, only to continue propagation at the next load cycle. Fatigue tests at varying frequencies would allow a fuller understanding of this type of failure but have yet to be carried out.

16.3.2 Loading conditions leading to fatigue failure

There are loading criteria which must be fulfilled if the fibers are to fail by fatigue. It seems likely that the fibers have to be subjected to a certain cyclic load amplitude for fatigue failure to occur although a minimum amplitude level has not been determined. However, what is clear in all of the fibers so far discussed is that the minimum cyclic load has to be below a certain level, but not necessarily zero, for fatigue failure to occur. Fig. 16.12 shows the effects of increasing the maximum cyclic load and also on increasing the minimum load on PET fibers subjected to fatigue loading at 50% at room temperature. It can be seen that increasing the maximum load, from 70% of simple tensile breaking load to 80%, reduces median lifetimes, as would be expected. However, if the maximum load is kept at 80%, increasing the minimum load from 0% to 10% of breaking load increases the median lifetime to the same as when the fiber is cyclically loaded from 0% to 70% of breaking load (Le Clerc et al., 2006a,b,c).

The loading levels shown in Fig. 16.12 are rather high and have been used so as to obtain failures in reasonable times. The fatigue process has been seen to be related to the internal damping, which occurs during cycling, and Fig. 16.13 shows how this energy dissipation is affected by changes of both the minimum and the maximum load levels (Le Clerc et al., 2007).

![Figure 16.12](image)

**Figure 16.12** The survival graphs of PET fibers subjected to different maximum and minimum cyclic loads. The median lifetime is defined as that which produces 50% survival rates.
It can be seen that, for any given loading condition, as the minimum stress is increased the dissipated energy falls quickly but as the maximum stress is reduced the dissipated reduces much less quickly. This suggests that although there is a minimum load cut-off level above which fatigue is inhibited, reducing the maximum stress only increases lifetimes but does not prevent ultimate fatigue failure.

Similar effects of changing loading parameters are also observed when testing PA fibers, as is shown in Fig. 16.14.

16.4 Mechanisms involved in fiber fatigue

The long fatigue cracks developed in the fibers that have so far been described is a reflection of the anisotropy of their molecular structures although the distinctive angles of penetration seen between the two main families of fibers, PET and PA, are not fully understood. It is possible that this angle is related to the long periods of the molecular morphology found in each type of fiber. The structures of these fibers are complex, as is shown schematically in Fig. 16.15. The structure of a PET fiber is thought to be very similar to that of PA fibers although perhaps showing greater groupings of the nanofibrils into larger fibrils, which may explain the longer breaks, observed in fatigue (Prevorsek et al., 1973; Marcellan, 2003; Herrera Ramirez, 2004).

These fibers are spun and drawn from the melt at very high speeds, 3000—7000 m/min. On leaving the spinneret the material is near its melting point, around 260°C and is quickly cooled. Cooling is most intense at the surface of the fiber, which is the region that first solidifies. Shortly afterwards the core of the fiber also cools and contracts.
This results in residual stresses across the fiber section with the surface being put into compression with respect to the core. The residual stresses have been measured by Raman spectroscopy and can be very significant (Marcellan et al., 2004). Cooling also produces a skin that can be observed by transmission optical microscopy and on scanning electron micrographs of fracture surfaces. At the molecular level the macromolecules are generally thought as being folded in compact crystalline regions, making the structure semicrystalline. The molecular structure is arranged in fibrils and possibly bundles of fibrils that must influence fatigue crack growth.

Figure 16.14 Energy dissipation during cyclic loading of PA 6.6 fibers, as the minimum and maximum stresses are varied.

Figure 16.15 Macro- and nanostructure of a PET or PA fiber (Prevorsek et al., 1973; Marcellan et al., 2004; Herrera Ramirez, 2004).
The fibers also contain materials other than the polymer. These are added to the polymer before extrusion and drawing for a variety of reasons. For example, antimony is added as a catalyst to aid polymerization in PET and bromine is added, often carried on flake glass, as an antioxidant or flame retardant material, in PA. These small inclusions are usually less than one micron in size but they are very significant in initiating crack growth in fatigue or possibly even under simple tensile or creep loading. Occasionally large particles can initiate failure from within the fibers when tested at room temperature. In this case the failure morphologies are conical, as can be seen in Fig. 16.16 (Herrera Ramirez and Bunsell, 2005, 2006).

One end of each fracture end can be seen to consist of a large cavity, which blends into the two regions of slow and rapid crack growth as seen in tensile and creep failures. The complimentary ends show a cone of material the apex of which can be seen not to be a point but rather a crater, in which, in some cases, a particle may be seen to be lodged. These fracture morphologies indicate that, unusually at room temperature, the fatigue cracks have been initiated at a point well under the fiber surface and that the slow crack growth at a slight angle to the fiber axis has produced the conical morphology, which, when the cross-section was sufficiently reduced, led to tensile failure. The advance of the fatigue crack front is shown by regular striations on the inside surface of the conical cavity.

Fig. 16.17 reveals a particle, which has been identified by SEM/EDS analysis as antimony, still in place in one of the two complimentary fracture surfaces of a PET fiber tested in fatigue but broken apparently in creep.

According to the above findings, the fatigue mechanisms initiated by a particle seem to be as shown in Fig. 16.18. The particle can be considered as an inclusion in the fiber, which provokes a weak adhesion between the particle and the fiber; this can lead to the presence of interfacial microcavity. Both the weak adhesion and the microcavity can locally cause an effect of stress concentration, which initiates the crack. The increase of the free volume in these interfacial regions contributes to the development of microporosity and the crack begins to propagate slowly and

Figure 16.16 Both broken ends of a PA 6.6 fiber broken at room temperature in fatigue at 50 Hz showing crack initiation inside the fiber. The small crater at the tip of the cone reveals the origin of the crack as an inclusion.
circumferentially with an angle towards the surface of the fiber. As the crack propagates, the cross-section of the fiber decreases until it no longer supports the applied load, causing a final rupture by a typical tensile process (zones of slow crack propagation and then rapid crack propagation). The particle can separate from the fiber at the time of the final rupture or remain at the bottom of the cavity.

An examination of the fibers before testing reveals the presence of the particles and there seems always to be a particle or several particles in the initiation regions of fatigue cracks (Le Clerc et al., 2006a,b,c). Fig. 16.19 shows the tip of the tongue obtained after the fatigue of a PA 6.6 fiber together with the complementary initiation point. The break has clearly been initiated just under the fiber surface by a particle.

Figs. 16.19 and 16.20 show that particles have initiated crack propagation in, respectively, a PA 6.6 fiber and a PET fiber, which have failed in fatigue. At room temperature, it is usually only the particles situated at the interface between the fiber skin and the core of the fiber, about one micron under the surface, which initiate cracks.
Figure 16.19 Complementary initiation points of a fatigue break of a PA 6.6 fiber showing that the initiation was by a particle, which has left a crater in the tongue end of the break (Herrera Ramirez and Bunsell, 2006).

Figure 16.20 The initiation of fatigue cracks, at room temperature, can be seen by transmission optical microscopy in PET fibers, to be associated with inclusions just under the surface probably at the interface between the skin and core of the fiber. The arrows indicate the particles and (d) shows how the stress field around the particle is modified.
Clearly the interface represents a weakened boundary within the fiber and the presence of these particles further weakens the fiber. The polymer, when drawn from the melt, undergoes considerable extension but the hard inclusions do not. This results in a region before and after each particle in which the polymer experiences different deformation from that of the rest of the fiber. This must create a weakened zone, which when it is at the skin-core boundary can initiate fatigue cracks and possibly other types of failure.

Fig. 16.21 shows two successive microtomed sections of the initiation region in a PA 6.6 fiber as observed by transmission optical microscopy. The presence of a particle at the skin-core interface initially induces debonding, which may be visible on the fiber surface by the presence of some irregularity. The skin is then broken and this is seen as the beginning of the longitudinal fatigue crack.

It should be noted that surface damage or irregularities producing associated stress concentrations are not the causes of the crack initiation but rather the disturbance, inside the fiber, due to the presence of a particle. When the fatigue crack has begun to propagate, its path can be influenced by other particles and it can be seen to be deviated so as to pass preferentially through regions near particles.

A study by Averett and Realff (2009) investigated the reinforcement of PET fibers by 5 wt% carbon nanotubes in monotonic and tensile fatigue tests at 5 Hz. They showed that compared to PET fibers that did not contain the nanotubes those that did showed poorer fatigue properties. This result seems to reinforce the damage initiation role of particulate inclusions in these fibers although the sizes of individual carbon nanotubes would be much smaller than those reported above. It is possible that with a 5 wt% content coalescence of the nanotubes occurred in these fibers.

Figure 16.21 Microtomed sections of the initiation region of a fatigue crack, obtained at 21°C and 50% rh, in a PA 6.6 fiber of 26 μm in diameter, revealing that initially the skin becomes separated from the core (left) and then the fracture breaks through the skin (right) to appear as the initiation point of the longitudinal fatigue crack.
16.5 Tensile and fatigue failure at elevated temperatures and in structures

The appearance of the fracture morphologies of both PA and PET fibers is seen to change as the temperature of the environment in which they are tested is increased. In tensile tests, the fracture morphologies become less crisp, more rounded, and sometimes more complex, particularly above the glass transition temperature of about 80°C, as can be seen in Fig. 16.22. The thickness of the modified zone due to the normal final fracture seems to be higher confirming the hypothesis of local melting, which is much easier with higher temperature of test condition.

Around the glass transition temperature ($T_g$) the fatigue fracture morphologies of both PA and PET fibers show two distinctive types of failure. The familiar long fatigue fractures found at room temperature can be found but increasingly, as the temperature is increased, another complex, truncated fatigue morphology, becomes the dominant feature of the fiber breaks, as shown in Fig. 16.23. Above $T_g$ only the truncated fatigue breaks are found.

These types of fracture morphologies are also found in fiber bundles that are cycled at 50 Hz at room temperature. The temperature of the bundles has been found to...
increase above $T_g$ due to the poor heat dissipation of the fibers so that even if the surrounding environment is at room temperature the fibers inside the bundle experience large increases in temperature and fail by the truncated fatigue process.

Fibers subjected to cyclic loading above $T_g$ are often found to fail with fibrillated fracture morphologies, as shown in Fig. 16.24. An examination of these complex breaks by transmission optical microscopy reveals that multiple breaks are initiated at high temperatures and throughout the body of the fiber. Fig. 16.25 shows such a
micrograph that again reveals that crack initiation is associated with the presence of particles. At these higher temperatures, however, the fractures can be initiated throughout the volume of the fiber.

**Fig. 16.26** shows a section of a PET fiber fatigued at 120°C and reveals an internal crack, which, as successive microtomed sections show, does not exit at the fiber surface (Le Clerc et al., 2007).

**Figure 16.25** Transmission optical micrograph of a PET fiber failed by fatigue at 80°C revealing multiple crack initiation points associated with particles distributed within the core of the fiber.

**Figure 16.26** Microtomed section through a PET fiber fatigued at 120°C showing two fatigue cracks that have initiated at the surface and an internal crack that does not exit to the surface.
The processes involved are an extension of the mechanisms seen at room temperature as initial damage is still seen at the skin-core interface but the presence of particles within the body of the fiber initiates additional crack propagation. It has been seen that these internal cracks do not necessarily exit at the fiber surface but clearly weaken the fiber. That the cracks do not reach the fiber surface means that there is no significant shear stress generated at the crack tip so that propagation along the fiber is limited. This explains the shorter conical breaks seen by Herrera Ramirez and Bunsell (2005, 2006). As has been demonstrated it is likely that several independent cracks can be initiated within any given length of fiber. Eventually these cracks coalesce and the complex truncated fatigue break occurs.

It seems likely that the increased temperature reduces transversal bonds between microfibrils making up the fibers so that the weak interface, provided by the skin-core boundary, is no longer unique and failure can be initiated throughout the body of the fiber encouraged by the presence of inclusions.

The truncated fatigue breaks observed in this study resemble exactly the fiber breaks taken from fatigued composite disk specimens consisting of the fibers embedded in a rubber matrix, as reported by Yabuki et al. (1986), Winkler (1991) and Naskar and Mukherjee (2004). It is clear that in this type of test, the fibers are subjected to large temperature increases due to internal damping and the poor heat transfer properties of the fibers and the rubber. The appearance of the fatigue breaks is then that of the truncated fatigue morphologies rather than those observed with single fibers tested at room temperature, for which heat exchange with the surrounding environment has been calculated to be of the order of only several degrees Celsius.

The simple tensile failure stress and strain to failure of the PET fibers tested at different temperatures vary as shown in Fig. 16.27.

![Figure 16.27](image)

**Figure 16.27** The failure strains and stresses of PET fibers as a function of temperature (Le Clerc et al., 2006a,b,c).
It can be seen that the failure stress is markedly different at 80°C and 120°C from that at room temperature. This can be of great importance if the fibers are thought to be at a lower temperature but because of internal heating, and poor heat dissipation, which may occur in fiber structures, they are really at a much higher temperature. In this case the fibers can be more susceptible to the fatigue process than would be the case if they were at lower temperatures. It can also be seen from Fig. 16.27 that the strain behavior of the fiber also varies with temperature.

These observations should be borne in mind when examining the fatigue results shown in Fig. 16.28 for which the percentages given are those of the maximum cyclic load compared to the simple tensile breaking load at that temperature. It should be noticed that the lifetimes decrease with temperature, although raising the minimum load is still found to increase lifetimes at any given temperature. The principal reason why lifetimes are reduced is that the final failure process in the fatigue of PET fibers is governed by the creep of the reduced load bearing section of the fiber. Creep is thermoactivated and its rate is increased with temperature.

Truncated fatigue failures have been observed at room temperature in PA 6.6 fibers that had been immersed in hot water and then tested at room temperature (Nasri et al., 2002). Significantly, these fibers absorb water and as they do the $T_g$ falls and can fall below room temperature. In this case it seems that the fibers were above their glass transition temperature and the fatigue failure resembled those obtained at high temperatures.

![Figure 16.28](image)

**Figure 16.28** Fatigue lifetimes of PET fibers as a function of temperature and stress level (Le Clerc et al., 2006a,b,c).
16.6 Conclusions

The most widely used and produced organic fibers, polyester and nylon fibers, fail by a tensile fatigue process, which produces a distinctive type of fracture morphology, very different from that obtained with other types of loading. Fatigue failures require sufficiently large cyclic load amplitudes but a necessary criterion is that the minimum cyclic load must be lower than a critical load threshold. The mechanisms that control this behavior are complex and the macrostructures of the fibers, which have been shown to have skin-core structures, are important in the initiation of fatigue breaks, at room temperature. The role of small, hard inclusions in the fibers has been shown to determine the point of initiation of the fatigue cracks which then run along the fiber, gradually reducing the load bearing cross-section until failure occurs. The angle of penetration of the fatigue cracks is common to the polyester fibers and seemingly independent of draw ratio. This angle is smaller than that seen with fatigued PA fibers and leads to longer fatigue cracks. At higher temperatures the crack initiation is found to occur at inclusions throughout the thickness of the fiber. This produces truncated failures similar to that found when the fibers in reinforced rubber are examined after disk fatigue tests.

Acknowledgments

This chapter refers to works in a number of research establishments and countries but could not have been written without the contributions of a considerable number of research students who have worked with the authors of this chapter. Some of these students are mentioned in the references but by no means all. Thanks must go to all of them for throwing light in a fascinating and difficult subject.

References


Handbook of Properties of Textile and Technical Fibres

Edited by A. R. Bunsell

The Handbook of Properties of Textile and Technical Fibres provides a thorough and authoritative overview of the properties of a wide range of natural and synthetic fibres.

Their presentation deals not only with their traditional textile uses but also shows how the remarkable properties of fibres can be exploited to allow the production of new and innovative structures. Natural fibres continue to find preferential applications in textiles, but increasingly some are also finding applications in engineering. Synthetic fibres find uses in textiles and also are allowing technical advances to be made, which would be impossible with more traditional materials. The unique properties of fibres have provided many crucial design solutions including in satellites, aircraft fuselages and wings, pressure vessels and mooring cables for off-shore oil-rigs and large ships. Fibre-based materials are now at the centre of an ongoing revolution in textiles, materials science and engineering.

Preliminary chapters give the reader an overview of the properties of fibres treated in the book as well as how their properties are measured, then separate chapters deal with the different classes and types of fibres.

The first part of the book examines the properties and structures of natural fibres, such as cotton, hemp, flax, agave, wool, cashmere, mohair, alpaca, and silk. Part two discusses the properties and structures of synthetic fibres ranging from polyamide, polyester, polypropylene, liquid crystal fibres, and high-modulus polyethylene fibres to glass, carbon, and ceramic fibres. Many chapters also provide a general background to the fibre, including their production, microstructure, and factors that affect their failure.

With its distinguished editor and array of international contributors, Handbook of Properties of Textile and Technical Fibres is an important reference for fibre scientists, textile technologists and engineers, as well as those in academia.

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