INVESTIGATION OF FATIGUE FRACTURE MECHANISM OF FIBRE REINFORCED INJECTION MOULDED POLYAMIDE

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In former work a correlation of fatigue performance with Ultimate Tensile Strength was found, this opens the possibility to predict the fatigue lifetime of products more accurately. In this paper the conditions for which this correlation is valid, is investigated.
For conditioned material a "bridged crack" mechanism was found: Damage starts with void initiation at fibre ends, the voids consequently grow along the fibre and merge into larger "cracks". However not one complete crack exists, as the crack walls remain connected at a number of spots.
For dry as moulded material the fatigue failure mechanism is different: The matrix material hardly shows any sign of ductility and no bridges are formed.

INTRODUCTION

Fibre reinforced injection moulded thermoplastics are different from materials like metals because of their high degree of anisotropy. The anisotropy is caused by fibre orientation, which is formed during injection moulding. Result of this is that tensile strength may vary between 100 and 160 MPa, within one product.

To be able to predict the fatigue lifetime of products more accurately, the correlation of fatigue performance with Ultimate Tensile Strength (UTS) was investigated. This correlation seems to imply that the UTS and the Fatigue stress are closely related, and that for predicting the fatigue behaviour of a product with a certain fibre orientation, the UTS can be used.

The reason for this correlation is unclear, therefore investigation of the mechanisms in both fatigue and tensile experiments has started. Apart from this, lifetime experiments on specimens under various conditions are executed, to be able to determine the factors that influence the beforementioned correlation.

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EXPERIMENTAL

The material used was Polyamide 6 containing 30%wt. of glassfibres; Akulon K224-G6 provided by DSM, the Netherlands. Square plates of 100x100 mm and of 2 and 5.75 mm thickness were injection moulded from this. Different qualities of the material were tested. These were caused by: 1) different conditioning: dry as moulded, conditioned to equilibrium water content at 23°C and 50%RH and wet. 2) Experiments were executed at higher temperatures (50°C) giving a different behaviour of the material. 3) A similar material was tested, which possessed improved fibre-matrix bonding. This was caused by a different fibre coating.

For the fatigue experiments non-standard dog-bone type test specimens were milled from the plate, using a Roland PNC-3000 Computer Aided Modelling Machine. The position of the different specimen types in the original square plate is shown as an insert in Figure 1. The width of the narrow part was 10mm.

The minimum to maximum load ratio was 0.1. Experiments were executed at a relatively low frequency of 1Hz. This because of the high heat build-up in the material, Horst(1,2). Especially when the material is conditioned to equilibrium water content, the damping is very high. Temperature during testing was 23°C and relative humidity was 50% for conditioned specimens, 30-40% for dry as moulded specimens. Wet specimens were tested in water, if low loads and consequently high lifetimes so required. Conditioned to equilibrium specimens were exposed to laboratory air for about 1 year.

SEM micrographs of the fracture surface were made using a JEOL JSM-840A after gold coating of the fracture surfaces in a Balzers SCD 040. To reveal the structure inside the specimen during the fatigue process, some specimens were first fatigued for a certain percentage of their expected lifetime, and consequently fractured after immersing them for 5 minutes in liquid nitrogen.

RESULTS

Some of the results of the lifetime tests for the 5.75 mm thick specimens are shown in Figure 1. Each point is the average for 5 experiments. When these Wöhler curves are normalized to the UTS in a tensile test, the data points for the different specimen types all fit the same curve (Figure 2). This normalization procedure was executed for specimens with 2 and 5.75 mm thickness, and for material that was dry as moulded and conditioned to equilibrium water content. Also the material with improved bonding was used. For each set the normalisation procedure results in a "master curve" as in

Figure 2. These master curves for the different material qualities tested are not the same however. Figure 3 shows the master curves for different qualities of material. In this graph it is easily seen that diverse qualities of material do not possess the same master curve. These differences are significant, as the scatter in lifetime amounts to a factor of two between the highest and lowest lifetime measured (0.3 on Log scale). The 2mm conditioned and 50°C specimens give the worst fatigue performance. The specimen with improved bonding and the one that was tested wet give the best (normalised) performance.

The 6mm and 2mm dry specimens, give an intermediate fatigue performance. These two master curves do coincide, proving that the material thickness is no important parameter in determining fatigue performance (the UTS of both specimens are equal as well).

SEM fractography was done on some specimens, showing some very interesting differences between the various material qualities. The conditioned and wet material showed high matrix plasticity on the fatigued fracture surface. The appearance of this is similar to plastically deformed parts of Figure 4. On the dry as moulded material the situation was entirely different. Matrix ductility is very small, leading to a flatter fracture surface. Of the conditioned material SEM fractographs were made of fatigued specimens (not up to failure) that were consequently emerged in liquid nitrogen and broken. As the cryogenic fracture is brittle, all ductility visible on the fracture surface must be caused by the fatigue process. In Figure 4 an example is given. Here a small brittle part (broken after emerging in liquid nitrogen) in a ductile area (fatigue) is visible.

DISCUSSION

The remarkable influence of water content on the composite ductility in a tensile test can be explained by the influence of the water on the fibre-matrix interface. As shown by van Hartingsveldt (3) for polyamide containing glass spheres, absorption of water lowers the strain at which debonding occurs. This detoriation of the interface causes the composite to behave more brittle, although the matrix material is considerable more ductile.

Fatigue lifetime at a certain stress level is directly related to the ultimate strength of specimens cut from different locations in a square plate. This indicates that a change in fibre orientation does not lead to a change in failure mechanism. Also the thickness of the specimens is not of importance, which is shown by the coinciding of the master curves for the 2 and 6 mm specimens in Figure 3. In other experiments a different fatigue behaviour of thick and thin specimens was observed. Measurement of fibre length showed a difference for both thicknesses, explaining the different fatigue behaviour in an indirect way.

The different qualities of material show the same effect, that Wöhler curves for different specimen types all fit the same curve. These master curves for different material qualities are not the same however. For example improving matrix fibre bonding does not improve the tensile strength, it does improve the fatigue behaviour though. It seems that also wet specimens give an improved fatigue behaviour. This is only the case because of the normalising procedure, the tensile strength is lowered (relatively) more than the fatigue strength. The conditioned specimen gives the worst fatigue behaviour. This can be explained: When an increasing amount of water is absorbed, first the fatigue strength is affected more than the tensile strength. When reaching satisfaction (100% water in the polyamide) the tensile strength is affected more than the fatigue strength, resulting in the relative improvement of the fatigue behaviour.

The brittle part seen in Figure 4 was broken during the cryogenic fracture, and therefore was a connection (bridge) between both crack walls. All observations and results can be explained by the "bridged crack" failure mechanism (for conditioned material and material tested at 50°C) Horst and Spoormaker(4), visualised in Figure 5: Damage initiates at the location of highest stress intensity, the fibre ends (1). Voids grow from this damage, mainly along the fibres (2). These voids and debonding of the fibres will largely reduce the reinforcing effect of the fibres, putting a higher stress on the matrix causing the matrix material to plastically deform in this damaged area. Increasing opening of this damaged area (3) will increase the straining of the material in front of the "craze" and will induce new voids and debonding, and growth of the "craze" in a direction perpendicular to the main stress direction. The voids continue to grow and merge into cracked areas (4). The cracked areas do not grow into one main crack though, the "crack" walls remain connected at certain locations. The mechanism for dry as moulded material is different, matrix ductility is much less, so no bridges can form, Lang et al.(5).

TABLE 1 Comparison of the Influence of various Factors on Tensile Strength, Fatigue Strength and the Normalised Fatigue Strength. "+" means that an influence exists.

factor:	Tensile strength	Fatigue strength	Normalised Fatigue str.
Fibre orientation	+	+	none
Fibre length	+	+	+
Fibre coating	none	+	+
Conditioning	+	+	+

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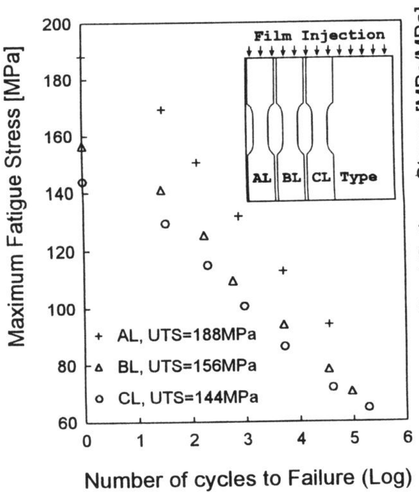

Figure 1 Wöhler curve for three types of dry specimens (see insert).

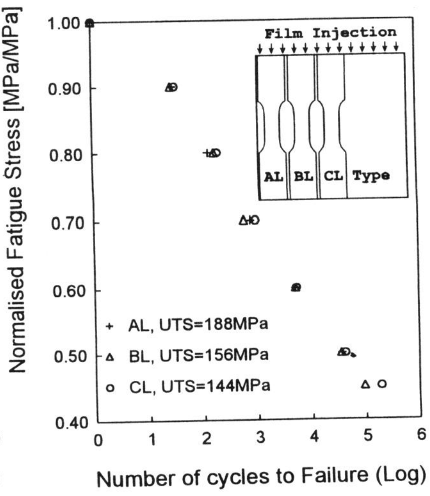

Figure 2 Normalised Wöhler curves for the same specimens as in Figure 1.

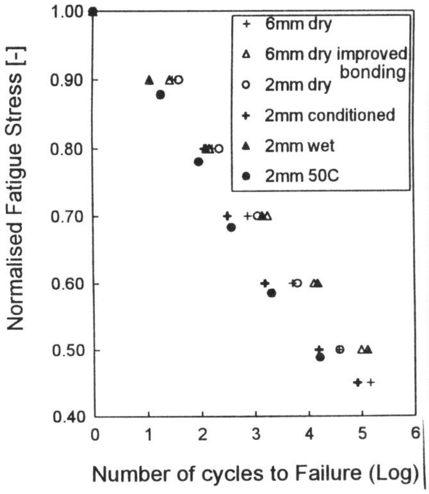

Figure 3 Normalised S-N curves for different material qualities.

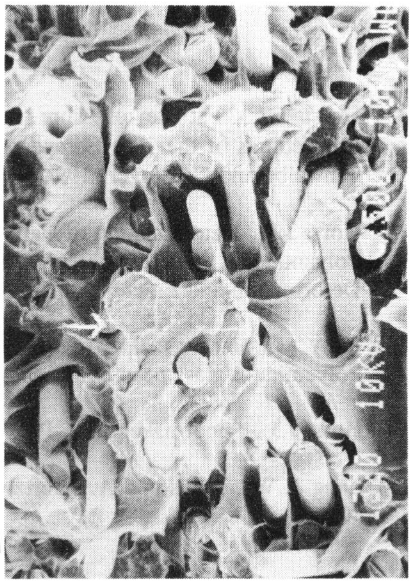

Figure 4 Fractograph showing ductile area and a (brittle) bridge (arrow).

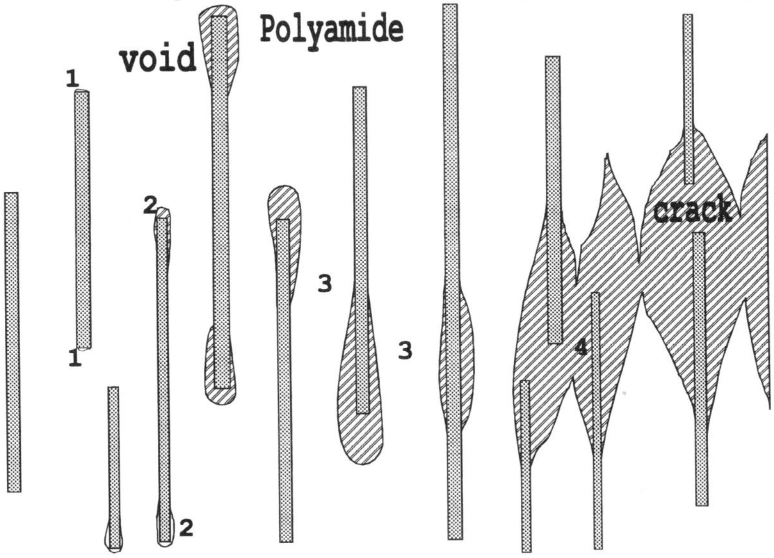

Figure 5 Visualisation of the fatigue failure mechanism for a model system, conditioned to equilibrium water content. Numbers refer to the text.