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FATIGUE OF CLOSED CELL FOAMS

Subtitle

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Abstract: This paper deals with fatigue of closed cell foams. The main idea is to use a few simple tests to predict the tension-tension fatigue properties of foams. The required testing consists of crack propagation rate measurements and one tension-tension fatigue test performed at yield stress for the foam. This data can then be combined to construct a synthetic S-N curve for the foam. Tests on three densities of Divinycell H-grade foam are performed and the results support this approach. Some preliminary results from two densities of Rohacell WF-grade are given as well. Static properties of foams scale with relative density and once this scaling can be obtained through various static tests and the same scaling appears to be valid for both crack propagation rates and fatigue properties of foams. The implication of this is that once the fatigue behaviour of one relative density foam is established, one can predict the fatigue behaviour of all other relative density foams within the same class of materials.

Key words: Cellular materials, fatigue, crack propagation, scaling

1. INTRODUCTION

Rigid cellular foams are used extensively as a structural core in load carrying sandwich structures. The usage stretches over applications in aerospace, automotive, marine, transportation and infra structure. There are numerous examples of applications and a few worth noticing here are for example the new Swedish Navy Corvette Visby, wind-mill blades, and novel train car structures. In all of these and most other foam core sandwich applications, the core is typically a closed cell polymer foam, designed to carry a substantial part of the load. The core in a sandwich structure has a

multitude of functions; carry transverse loads (through shear) and distribute load between the face sheets, stabilise the face sheets to avoid instability, transfer local loads acting on the face sheets or via fasteners, etc. The core usually also provides a number of other functions, such as thermal insulation, acoustic insulation, etc. More and more has been focused on the core material recently due to increased demands for material properties and models to use in the design of sandwich structures. Fracture and fatigue of load carrying foam cores still remains to a large extent unknown. The reason for this is the inherent structure of foams, constituted of a complicated 3-dimensional network of thin membranes (cell walls), enclosing each cell. At the intersection of cell walls, edges with concentrated mass build up rods or beams. A foam is thus not really a material, but a micro-structure. The macroscopic properties of this structure (material) depend on many parameters and not only the properties of the material building up these walls and edges. One important property is of course the relative density, $\bar{\rho}$, defined as the ratio of the density of the foam ρ divided by the density of the solid material ρ_s . The cell structure itself will also strongly affect the macroscopic properties of the foam. Yet another important parameter is the amount of solid material that builds up the cell walls and edges, respectively. It is well known that open cell foams (with all the solid material is in edges) behave differently than closed cell foams.

In the well-known textbook by Gibson and Ashby [1] cellular foams (3D cellular structure materials) are modelled using the relative density as the primary variable. A variety of properties are modelled using the relative density, the properties of the solid cell edge/face material and the cell structure. Rather straight forward kinematic relations are used and very useful, simple formulae for design are derived. In essence, most properties of foams can be written as

$$\bar{x} = \alpha \bar{\rho}^n \quad (1)$$

where x is some mechanical property of the foam normalised with its value for the fully dense material (bulk property) of which the cell edges and faces are made of, in the case of polymer foam thus the solid polymer material property. Herein, we shall assume that the properties of the foam can be described by means of eq.(1), where the exponent n will be found from fitting experimental data. However, this exponent must have a value $1 \leq n \leq 2$ [1].

2. MATERIALS

Two closed cell foams were used in this study; Divinycell H-grade and Rohacell WF-grade. Both have a closed cell structure and belong to the category of high-performance rigid polymer foams used as load carrying core materials in various applications

Divinycell is a cross-linked rigid cellular PVC foam and it is produced in a variety of densities where mechanical properties (higher strength and moduli) increase with density. Even a small variation in density will cause differences in the mechanical properties. The fracture behaviour of this material is in the regime between ductile and brittle. Divinycell qualities used in this survey were H60, H100 and H200, where H stands for the grade and the number corresponds to the nominal density in kg/m^3 . Any details on this material can be found in [2].

The other material used in this study is Rohacell, a PMI foam with predominantly closed cells and which is more brittle than the PVC foam. The qualities used herein were WF51 and WF200, where WF is the particular grade of Rohacell and the number corresponds to the nominal density in kg/m^3 . Details on this material can be found [3].

Both foams have similar micro-structure with most of the solid material in the faces or cell walls. There is a portion of material concentration at the edges, the junctions where the cell walls meet. One could assume that these materials then should behave close to that of a perfect closed cell structure exhibiting a cell wall stretching deformation mechanism. However, the concentration of mass in the junction has some effect. By fitting data from tests performed in-house on various densities and also using data supplied by the manufacturers the relations between properties and relative density was adapted to an equation of the type in eq.(1). The values for ρ_s and E_s are simply estimates using handbook data for similar polymeric materials. It was found to be difficult to obtain bulk properties for these polymer blends. Changing these bulk data will imply changes in the curve fitting parameters, though the slope n will not change presumably. Different testing methods give slightly different results, but in short it was found that the slope according to eq.(1) varied between 1.1 and 1.2 for all static mechanical properties of Divinycell H (elastic modulus, ultimate tensile and compressive strength, shear strength, and fracture toughness). A value of $n = 1.1$ gives quite decent curve fits to all tensile properties. The exponent number for Rohacell WF appear also to be around $n = 1.1$. These are the values used in subsequent scaling of test results.

3. STATIC FOAM PROPERTIES

A number of static tests were performed on the materials. They were preformed using the same test up as in the tension-tension fatigue testing, which is described below. Fig.1a shows tensile stress-strain curves for H60, H100 and H200. The same curves are plotted again in Fig.1b, where the stress is normalised with the relative ratio so that $\bar{\sigma} = \sigma / \bar{\rho}^n$. As seen in Fig.1b., the three stress-strain responses for three different relative densities collapse into a single curve. A similar normalisation with results from Rohacell WF materials gives similar results.

Figure 1 (a) Typical stress-strain relations for H60, H100 and H200 (from bottom to top) and (b) normalised stress-strain relations.

From this testing, all the necessary material characteristics used in this investigation can be extracted, including the Young's modulus, E , and the yield strength σ_y , here defined as the stress at 0.2% offset strain, as commonly defined for other engineering materials which does not exhibit a distinct yield point. The material data for the materials used herein are given in Table 1.

	H60	H100	H200	WF51	WF200
E [MPa]	78	126	304	89	n.a.
$\sigma_{0.2}$ [MPa]	1.31	2.37	4.61	1.46	n.a.
σ_{ult} [MPa]	1.83	3.34	7.14	1.60	n.a.
K_{Ic} [MPa $\sqrt{m}$]	0.13	0.21	0.41	0.10	0.34
C [*]	2.44	0.139	1.87E-3	2.01E+11	2.69E+3
m	5.68	5.91	5.48	13.1	13.5

Table 1 Material data for Divinycell H and Rohacell WF materials (* dimension obtained using da/dN in m/cycle in eq.(1))

4. CRACK PROPAGATION RATE TESTING

Before moving on to fatigue testing, a short review of some earlier tests results is necessary. In [4] crack propagation tests of Divinycell H and Rohacell WF were performed using a modified version of the compact tension (CT) specimen. The crack propagation measurements were fitted using Paris' law, which reads

$$\frac{da}{dN} = C \Delta K_I^m \text{ where } da/dN \text{ is given in m/cycles} \quad (1)$$

There are few important findings from this investigation; the crack growth rates yield very high Paris' law exponents $m = 5.5-6$ for Divinycell H and as large as $m = 13$ for Rohacell WF. The numbers for C and m for the material used herein are given in Table 1. It was previously shown by Huang and Lin [5] that fatigue data for open cell foams can normalized with relative density.

The Paris' law curves for Divinycell H60, H100 and H200 are shown in Fig.2a. The same type of normalisation can again be done by normalising ΔK with the relative density using $\Delta K/\rho^n$ and if this is done the curves will collide into one single relation, as shown in Fig.2b.

Figure 2 (a) da/dN vs. ΔK curves for Divinycell H60, H100 and H200 (left to right) and (b) the same curves but for a density normalised ΔK .

As seen in Fig.2b, the Paris' law relation also appears to be generic for a class of foams. In short, this implies that once the property – relative density relation is known, one test should be ample to know the properties for all densities of the same material. In this case, the property – relative density relation can be found by performing static Mode I fracture toughness tests of various foam densities. To create one master curve for all da/dN data we can find the slope in Fig.2b, which is around $m = 6$ and the corresponding number for C in the normalized graph. This can then be extracted for all relative densities and used for predictions.

5. TENSION-TENSION FATIGUE TESTING

The fatigue test procedure used an axi-symmetric dog bone specimen, described in ASTM D1623-78 “Tensile and tensile adhesion properties of rigid cellular plastics” [6]. The test fixture consists of two similar aluminium cylinders with one flat end to which the test specimen was bonded and one threaded end that was fixed in a hydraulic test machine. The specimens were cut from blocks of foam core, bonded between the two cylinders and a waist was milled to the correct shape and size in a lathe. The two cylinders were threaded in opposite directions in order to simplify the fastening into the test machine. The test rig with a specimen is shown in Fig.3.

Figure 3 Schematic drawing of the test set-up with a temperature sensor mounted on the specimen. Dimension used $D = 50$ mm, $d = 30$ mm, $H = 50$ mm, and $h = 32$ mm.

Static tests were performed in tension under a constant deformation rate of 2 mm/min at room temperature using the same specimen type. The fatigue

tests were performed under a load controlled sinusoidal cycle using a servo hydraulic testing machine, MTS 50kN. The load ratio used was in all cases $R = 0.1$, $R = \sigma_{min}/\sigma_{max}$ and a testing frequency of 5 Hz. The fatigue life of the specimens is characterised as the number of cycles to ultimate failure.

6. STRESS-LIFE AND PARIS' LAW CORRELATION

The analytical background is based on converting the classical Paris' law to a Basquin's type expression by calculating an inherent initial flaw size. This initial flaw size appears to be the same for all densities of a particular class of foams. The flaw size can be obtained from one fatigue test performed at the yield stress, and is thus not very time consuming.

By assuming crack growth in closed cell polymer foam to be continuous, as in a continuum, we can use the measured crack growth data in a similar way to what commonly is done for metals. Assuming the foam to behave like a continuum implies that there is no evident length scale present for cracked bodies. The approach to find a link between the S-N curve and the crack growth measurements is then similar to that of homogeneous materials. We take the S-N curve to be a crack growth test from some *initial flaw size* a_0 to a final crack size a_f . The initial flaw size is then considered to be some typical defect present in the volume of the material and the final crack size is simply some defect size in the test specimen which causes fracture at the maximum stress in the load cycle. Starting from Paris' law we have

$$\frac{da}{dN} = C \Delta K_I^m = C \left(\Delta \sigma_0 \sqrt{\pi a} \right)^m = C \Delta \sigma_0^m (\pi a)^{m/2} \quad (2)$$

with experimentally known coefficients. We can integrate this to find the number of cycles required to propagate a crack of some initial length a_0 , usually denoted the *initial flaw size*, to a critical length a_f . The critical length is simply given by some length that leads to static failure at peak load during a fatigue cycle. This is given by

$$a_f = \frac{1}{\pi} \left(\frac{K_{Ic}}{\Delta \sigma_{0,max}} \right)^2 = \frac{1}{\pi} \left(\frac{K_{Ic}}{\Delta \sigma_0 / (1-R)} \right)^2 \quad (3)$$

The number of cycles to failure is then

$$\int_0^{N_f} dN = \int_{a_0}^{a_f} \frac{da}{C\Delta\sigma_0^m (\pi a)^{m/2}} \quad (4)$$

The number of load cycles to failure, N_f , is then given by

$$N_f = \frac{1}{C\Delta\sigma_0^m (\pi)^{m/2}} \left(\frac{(a_f)^{1-m/2} - (a_0)^{1-m/2}}{1-m/2} \right) \quad (5)$$

This is obviously simplified to the point that the stress intensity factor has no finite width correction factor and only depends on the square root of the actual crack length. For simplicity, let us also first assume that the final crack size is much larger than the initial one, so that the fatigue life of a small specimen is the same as for a large one. The first reason being that the exponent in Paris' law is quite high, for example $m = 6$ for Divinycell and approximately 13 for Rohacell. The second reason is that the critical flaw size comes out as a rather small number, using the fracture toughness of the materials used. If so, the number of cycles to failure will be

$$N_f \approx \frac{1}{C\Delta\sigma_0^m (\pi)^{m/2}} \left(\frac{(a_0)^{1-m/2}}{m/2-1} \right) \quad (6)$$

For example, if we consider a sheet of H100 with a central crack. The critical crack length can be evaluated by means of the following; If $\sigma_{0,\max}$ equals the yield strength (2.37 MPa) and the fracture toughness is approximately $0.21 \text{ MPa}\sqrt{\text{m}}$, the critical flaw size a_f becomes only 2.5 mm. This is commonly much smaller than typical test specimen dimensions. Eq.(6) can now be rewritten to a Basquin's law type equation

$$\Delta\sigma_0 \approx \frac{(a_0)^{1/m-1/2}}{C^{1/m} \sqrt{\pi}} (m/2-1)^{-1/m} N_f^{-1/m} \quad (7)$$

or

$$\sigma_{0,\max} \approx \frac{(1-R)(a_0)^{1/m-1/2}}{C^{1/m} \sqrt{\pi}} (m/2-1)^{-1/m} N_f^{-1/m} \quad (8)$$

where $\sigma_{0,\max}$ is the maximum stress. The slope of this curve is $-1/m$ and is thus linked to the slope of the Paris' law curve. This equation has limited validity, i.e., we go from high-cycle fatigue described by Basquin's law, to low-cycle fatigue when the maximum applied stress exceeds the yield strength σ_y , here defined as $\sigma_{0.2}$. This occurs when

$$N_{f,\text{limit}} \approx \frac{(1-R)^m}{C\sigma_{0.2}^m(\pi)^{m/2}} \left(\frac{(a_0)^{1-m/2}}{m/2-1} \right) \quad (9)$$

Now we run into some problems since both $N_{f,\text{limit}}$ and a_0 are unknown. The way to proceed proposed in here is to perform at least one tension-tension fatigue test at which $\Delta\sigma_{0,\max} = \sigma_{0.2}$ to get an estimate for $N_{f,\text{limit}}$. Once this number is established, the initial flaw size can be calculated. This is done by converting eq.(9) to

$$a_0 \approx N_f^{\frac{1}{1-m/2}} (m/2-1)^{\frac{1}{1-m/2}} C^{\frac{1}{1-m/2}} \Delta\sigma_0^{\frac{m}{1-m/2}} (\pi)^{\frac{m}{2(1-m/2)}} \quad (10)$$

We can get the initial flaw size by means of knowing the Paris' law constant C , the Paris' law exponent m and one point on the S-N curve. This point is chosen as the number of cycles to failure when the maximum stress equals the yield strength of the foam and thus requires testing only up to a finite number of cycles. This S-N curve is then valid for cycle numbers in the high-cycle fatigue regime, i.e., down to $N_{f,\text{limit}}$.

7. RESULTS AND DISCUSSION

The results from the tension-tension fatigue testing Divinycell H60, H100 and H200 are presented in Fig.4(a) as a log stress vs. log number of cycles to failure. The horizontal line corresponds to the yield stress ($\sigma_{0.2}$) of the material. In order to get the synthetic S-N curve for the material the following data has been used; firstly, the number of cycles to failure for the specimens tested at a maximum stress $\sigma_{\max} = \sigma_{0.2}$ has been used find the initial flaw size, a_0 , through eq.(10). For Divinycell, the number of cycles to failure at yield stress is approximately 10000, as seen from Figs.4(a), but this is a lower limit. This gives an initial flaw size in order of 0.2 mm. For Rohacell the corresponding transition from low-cycle to high-cycle fatigue is around 100 cycles as a conservative estimate and this gives an initial flaw size of approximately 0.3 mm.

Fig.4b shows the results re-plotted by normalising the stress with the relative density. The relative density normalised S-N curves appear to collapse into one single master curve. The slightly different slopes on the predicted S-N curve (straight lines in Fig.4) are due to that the Paris' law exponents are slightly different for the different densities.

Figure 4 S-N curves for (a) H60, H100 and H200. (b) Stress normalised S-N curves for all materials.

8. CONCLUSIONS AND FUTURE WORK

The main aim with this investigation was to use Paris' law data to predict the fatigue life of foams under tension loading. Paris law data for the foams were used together with one tension-tension fatigue test to predict the S-N curve for the foam. The correlation between this prediction and the experimental obtained fatigue data is quite satisfactory. It is further seen that static properties of a certain class of foams can be normalised with the relative density and a unique, density independent property can be obtained. The same normalisation can also be performed on Paris law relations and fatigue data. The conclusion of this is that the fatigue life in tension for one class of foams appears to be possible using one (or perhaps a few) tension-tension fatigue tests at e.g. yield stress and the crack propagation data for one foam quality. These tests are both fairly fast and cheap to perform. Fatigue data can then be predicted for other densities using relative density normalisation. Although only test results for Divinycell H-grade foams are presented herein, the same approach appears feasible for Rohacell WF-grade foams.

The long-term aim of this project is to be able to predict the fatigue life of foams, under general loading conditions, without the need for extensive testing. This could be particularly important in the design process of

structures. Secondly, gaining certainty into issues like fatigue, safety factors could potentially be decreased thus saving structural weight.

The next problem treated is fatigue under compression loading. Fatigue tests will be performed at reversed loading with reference to the compressive strength of the foam. The reason for this is the correlation with shear loading [7], being a combination of tension and compression. From this, comparisons can be made with fatigue test results from beams in which the core is mainly subjected to shear fatigue loading.

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