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The outcome of our research can be utilized by metal foam manufacturers to improve their product. An example of this is Recemat International, a producer of nickel-chromium open-cell foam in the Netherlands. It is a supplier of foam parts for a rotating trickle-bed reactor that find application in the Fisher–Tropsch synthesis of for example, wax, liquid-phase hydridesulfurization of kerosene and naphtha. In fact, metal foam in a rotating trickle-bed reactor enhances the reaction rates up to 30-40 times compared to conventional trickle-bed designs [1] and it decreases the reactor size from 60 m$^3$ to 1.5 m$^3$. Centrifugal forces and drag of the fluids and gas impose a considerable load on the open-cell foam parts, i.e. increasing with radius, mass and rotational speed. Clearly, the rotation speed and radius are limited by the strength of the foam and the centrifugal load should not exceed the UTS. The fatigue life of the foam is also affected by the load during service and it decreases with increasing load. On the other hand production throughput increases with the radius and the rotation speed. To optimize the fatigue life and throughput the radius and rotation speed should be optimally chosen. Detailed knowledge of the mechanical performance of the foam contributes to the optimization of the design.

Recent progress in processing routes presents new challenges to the understanding of intrinsic metal foam behavior that require further investigations. In this section a number of outstanding issues and possible future research directions are proposed. These opportunities will contribute to a more quantitative and mechanism-based understanding of the mechanical behavior of metal foam.

Density variations

When we assume that the foam is a power-law hardening material and the exponent is independent of the density (Chapter 7), the strength of the foam and therefore the proportionality constant, $k$, depends on the relative density, $\rho_r$:

$$\sigma^* = k(\rho_r)\varepsilon^n$$  \hspace{1cm} (O.1)
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When a density variation around a mean value is present along the gage length of a sample the strain of the whole sample, $\varepsilon$, at a certain stress, is the average of the strains of the individual cross-sections with their specific density:

$$\bar{\varepsilon} = \sum_{i=1}^{m} \frac{1}{m} \varepsilon_i$$  \hspace{1cm} (O.2)

Substituting Eq. 1 into Eq. 2 results in:

$$\sigma^* = K_{tot} \bar{\varepsilon}^n$$ \hspace{1cm} (O.3a)

where

$$K_{tot} = \left( \sum_{i=1}^{m} \frac{k_i}{n} \right)^{-n}$$ \hspace{1cm} (O.3b)

The proportionality constant, $k_i$, depends on the density and the proportionality constant of the mean density, $k_{\bar{\rho}}$, and respectively, $\rho_r$:

$$k_i = k_{\bar{\rho}} \left( \frac{\rho_i}{\rho_r} \right)^{h}$$ \hspace{1cm} (O.4)

Substituting Eq. 4 into Eq. 3 leads to:

$$K_{tot} = k_{\bar{\rho}} \left( \sum_{i=1}^{m} \frac{1}{m} \left( \frac{\rho_i}{\rho_r} \right)^{\frac{h}{n}} \right)^{-n}$$ \hspace{1cm} (O.5)

This shows that for a power-law hardening material with a density variation the proportionality constant of the sample, $K_{tot}$, is not the same as that of the mean density of the sample, $k_{\bar{\rho}}$. For a Gaussian distribution of the density variation with a standard deviation of 10 percent of the average and a strain hardening exponent of the foam $n=0.28$ the values for the term in brackets in Eq. 5 are respectively 0.95, 0.94 and 0.78 for values of $h=1.5$, 1.63 and 2.7. This could explain why two samples with the same density, but probably not the same density variations, can have different strengths.
Hardening

The fact that the strain hardening exponent of the foam is not identical to the strain hardening exponent of the base material for the HT samples and the linear hardening behavior of the foam at larger strains indicate that not only intrinsic base material hardening is present, but also another source of hardening is present. It is likely that this second source of hardening originates from the change in topology, which we have been referring to as “geometric hardening”.

Damage

The uniaxial failure process of metal foams differs from failure of a solid metal. In ductile solid metals the stress is rather uniform until it localizes into a neck at the UTS. In metal foams, however, due to the stochastic nature of the cellular structure, a non-uniform stress and strain distribution is always present (see Chapter 9 and Fig. O.1a). The strain distribution depends on the density variation (see Chapter 9) along the gage length of the sample. Within each cross-section a stress distribution is present among the struts in that cross-section. This is due to different strut shape, thickness and orientation and possibly due to density variations within the cross-section and the interaction with neighboring cross-sections.

Macroscopic failure of the foam occurs through the consecutive fracture of individual struts, which, due to the ductility of the strut material, does not occur instantaneously after fracture of the first strut, but develops more gradually as a function of strain (see Fig. O.1b). It is estimated that individual struts under a high stress state fracture, but the mean stress still increases due to geometric and intrinsic base material hardening in the remaining struts. The peak stress is reached at the instant that for a small increase in strain the reduction in load-bearing capacity due to failing struts (i.e. damage) becomes larger than the total increase in load in the remaining (still intact) struts. Therefore, a non-uniform stress distribution is the cause for the introduction of damage before the peak. The additional strain the material can bear after failure of the first strut depends critically on the ability of the remaining struts to handle the additional, redistributed stresses of the failing struts. The number of struts that can fail in a cross-section before the peak stress is reached is a measure for the tolerance of the foam to damage.

When all the three abovementioned subjects, i.e. density variations, hardening and damage are known, in principle it should be possible to predict the complete stress-strain curve in tension of each sample with reasonable accuracy. So far, in Chapter 8 only one particular point in the compressive stress-strain curve, i.e. the
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![Figure O.1 Schematic overview of the fracture process of metal foam. (a) $\sigma$ is the stress, in the loading direction, on the struts in one particular cross-section. (b) $\sigma^*$ is the mean stress of the foam. The numbers in (a) correspond to the numbers in (b) and the black area in the stress distribution (#4) represents fractured struts. Changes in strut properties and loading due to a change in density or in base material properties affect either the stress distribution and/or the fracture stress distribution. The peak stress is reached at the instant that for a small increase in strain the reduction in load-bearing capacity due to failing struts (i.e. damage) becomes larger than the total increase in load in the remaining (still intact) struts. (b) Schematic of the stress-strain curve and associated damage evolution. The dashed line represents the stress-strain curve without the influence of damage.](image)

plastic collapse stress, has been predicted with reasonable accuracy, but there was still a lot of scatter amongst the samples and tensile damage by means of broken struts was not present.

The stress-strain relation in tension would become something like this:

$$\sigma = C\varepsilon^n \cdot F(\text{hardening, damage})$$  \hspace{1cm} (O.6)
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Where C is a constant that depends on the proportionality constant of the topology, the yield stress of the solid, the orientation of the long axis of the cells, the average density and the density distribution of the sample. This should give the correct yield stress of the foam without scatter between the samples. The evolution of the stress as a function of strain should be described by $\epsilon^p$ and $F(\text{hardening, damage})$. $\epsilon^p$ gives the intrinsic base material hardening and the function, $F(\text{hardening, damage})$, describes the geometric hardening and damage accumulation part of the sample. The damage accumulation can be measured by means of electrical resistance or X-ray tomography measurements.

Another subject worth researching is the difference in damage accumulation between open- and closed-cell foams. It has been shown in Chapters 6 and 7 that when the stretching contribution on the struts decreases the damage accumulation decreases. However, for closed-cell foams the stretching contribution in the cell wall is very large. It is therefore expected that the damage accumulation in closed-cell foam is much larger than in open-cell foams. Measurements that confirm this suggestion can be found in Fig. 2 of Chapter 3 and in Fig. 2 of Ref. [1] of Chapter 7. Fig. 2 of Chapter 3 shows no hardening behavior at all, the peak stress immediately follows after the elastic behavior. This indicates that damage accumulation is fast, which is possible by the microstructure of the particular foam. In Fig. 2 of Ref. [1] of Chapter 7 it can be observed that the presence of 10% brittle Si and high stretching in the cell wall results in a stress-strain curve with no hardening at all. Although the material is stronger than the foam with 0.6% Si the peak strain and stress are lower due to the fast damage accumulation.

The presence of brittle phases in the microstructure of closed-cell foam is very important for the monotonic fracture behavior as indicated above. It can also have a very large effect on the fatigue behavior when the cyclic loads are too high and damage is introduced during the first cycles. A sharp crack tip can propagate during the rest of the test and leads to early failure of the sample. Also in fatigue there will be large differences between open- and closed-cell foams. A sharp crack tip, which is present in closed-cell foams, is not present in open-cell foams due to the missing cell walls. How this affects the fatigue behavior of open-cell foams is unknown. When the cyclic loads are not high enough to introduce damage cracks can initiate in the aluminium dendrites (see Chapter 4) and not in the secondary phases. These cracks initiate at locally thinner part of the cell wall. The cell wall thickness increases with the cell size and therefore it could be possible that the initiation of fatigue cracks is postponed with increasing cell size.
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The ability of the low density replicated foams to shift the deformation from a highly strained band to another one is an unexpected result of the research performed in this thesis. It is a combined result of the tolerance of metal foams to damage and their ability to plastically deform and alter the topology. The behavior has not been observed in local strain measurements of the LD Duocel foam samples for different heat treatments and relative densities. It has only been directly observed for replicated foams and the number of bands involved has been related to the relative resistance measured at the peak strain. The relative resistance at the peak strain of the Duocel foam samples is nearly constant for all the samples, except for a few medium density TD samples. The relative resistance at the peak strain in those samples is about twice as high and both the q-value and the proportionality constant of the topology, $C_E$, of those samples is the same as for the replicated foam samples (see Chapter 7 and 9). This indicates that it could be possible that multiple bands were involved in the fracture process for these medium density TD Duocel samples, however the local strains of these samples were not measured and therefore this can not be verified. It is an interesting subject for future research.

References