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Properties of heat-treated aluminium foams

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Abstract - Aluminium foams based on the wrought alloy AA6061 were subjected to different heat treatments in order to evaluate possibilities to tailor their mechanical properties. Quenching after foaming was carried out either in air or in water. Solution heat treatment conditions were varied from "no treatment" to treatments combined with different air and water quenching methods. This gave rise to different cooling rates which were measured in-situ. Full age-hardening cycles were compared to heat treatments in which a simplified programme was applied. In total, the effect of nine different heat treatments on micro-hardness and the compression properties of foams was investigated. Foams which were fully heat treated performed best with an increase of strength by 60-75% over the untreated samples. However, age-hardening directly after foaming also caused significant improvements of strength, thus allowing to omit water quenching steps which carry the danger of damaging the foamed structure.

Keywords - Aluminium; Foam; Heat treatment; Processing

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1 Introduction

Metal foams are a relatively new class of structural materials and offer a variety of applications in fields such as lightweight construction or crash energy management. Recent improvements as well as new developments of production methods for metallic foams have given rise to renewed interest in these materials [1,2]. Currently, aluminium alloy foams are on the verge of being used in industrial applications.

One manufacturing method using metal powders is especially suited for the production of net-shape parts. It starts from mixing metal powders with a blowing agent, after which the mix is compacted to a dense pre-product. The resulting compacts are foamed by melting, thus triggering the release of gas from the blowing agent and the corresponding formation of bubbles. Detailed descriptions of the process are given in the literature [3-5]

In view of potential applications the mechanical properties of foamed metals are of paramount interest. It has been shown that, e.g., compression strength is connected to the density of a foam [1,6], thus allowing to adjust this property within a certain range. However, because density cannot always be varied freely and in order to gain more control over the properties of metallic foams, adjustment of other variables seems desirable, namely of alloy composition, foam morphology (size and shape of cells) and the metallurgical state of the matrix metal. The purpose of this paper is to evaluate the effect of the latter, i.e. the influence of different heat treatments on the structure and mechanical properties of aluminium foams. Only sparse results on the heat treatment of metal foams have been published in the past [6-9]

The 6XXX series of wrought aluminium alloys is widely used in automotive and food industries and for architecture, e.g. in trucks, for a variety of interior parts in cars, in railway carriages, pipelines or furniture. 6XXX alloys are chosen whenever heat-treatable alloys combining medium strength with good corrosion resistance, reasonable weldability and moderate cost are required [10]. The alloy AlMg1SiCu (AA6061) was selected for this reason and also because of its susceptibility to heat treatment and its well-investigated foaming behaviour. Moreover, AA6061 was also intended to serve as a prototype for age-hardenable alloys. Results of general nature are expected which would be transferable to other alloys.

Most heat treatments aim at controlling strength and ductility. For AlMgSi- and related alloys, maximum strength is achieved by means of precipitation hardening at the cost of ductility. Ductility can be increased via annealing – at the cost of strength. In order to cover the entire range of heat treatments the investigations included complete precipitation hardening cycles as well as pure annealing treatments. Furthermore, the possibility of omitting the solution heat treatment during precipitation hardening was studied, assuming that a supersaturated solution of precipitate-forming elements might already be achieved during cooling after foaming. A well-defined processing of the specimens was ascertained by measurement of the temperature curves for critical steps within the heat treatment cycles. After heat treatment the samples were characterised by metallography, EDX analysis, micro-hardness measurements and compression tests.

2 Heat treatment of AA6061 alloys and foams

2.1 General considerations

6XXX alloys contain Si and Mg as main alloying elements with other elements such as Cu and Mn for improving mechanical properties. The physical process effective in heat treating 6XXX alloys is *precipitation hardening* and can be divided into three steps: During *solution heat treatment* (SHT) at a temperature T_s for a period t_s a solid solution of the alloying elements is formed. *Quenching* at a rate $(dT/dt)_q$ allows for stabilising this state at room temperature thus leading to a supersaturated solid solution. After an optional delay time t_d, *ageing* at temperature T_a for a period t_a leads to the formation of precipitates in fine dispersion within the supersaturated regions [11,12]. Table 1 gives numerical values for the relevant parameters applicable to AA6061 alloys as found in the literature.

The aim of heat treatment is to produce a uniform distribution of coherent or semi-coherent precipitates while avoiding development of larger equilibrium structures without coherence with the matrix lattice. An impropriate selection of parameters will yield inferior microstructures. One source of problems is insufficient quenching. Especially in the critical temperature range between 400°C and 290°C the probability of unwanted precipitate formation is highest if the alloy passes through this range too slowly. For this reason, the average cooling rate in this temperature range, $(dT/dt)_q$, should in theory exceed 100 K/s for alloy AA6061 [10] to achieve a perfect solid solution. Quench rates below this limit will result in a higher fraction of undesired grain boundary precipitates. In technological heat treatments, this effect is accepted in many cases, e.g. when achieving maximum quench rates requires special, costly measures, carries the danger of warping or is simply impossible because of the sheer volume of components.

In contrast to precipitation hardening, *annealing* is meant to increase ductility, to eliminate the effects of strain hardening in cold-worked components or of prior heat treatments, and to avoid any hardening effects through natural ageing. A coarse and widely spaced distribution of precipitates is desirable, while on the other hand for AlMgSi(Cu) alloys temperatures of 415°C should not be exceeded due to possible oxidation and grain growth. After a soak time $t_{an,1}$ at a temperature $T_{an,1}$ subsequent slow cooling at $(dT/dt)_{an,1-2}$ down to $T_{an,2}$ accounts for high diffusion rates of precipitate forming elements and thus results in maximum coalescence of precipitates. Table 1 lists some practical values for the alloy under consideration.

2.2 Heat treatment of foams

It is not obvious that the procedures and parameters for heat treating bulk aluminium alloys can be transferred directly to corresponding foamed alloys. There are several reasons for this:

(i) Foams are stable only because the matrix metal contains finely dispersed oxide particles originating from the former surfaces of the powder particles the foamable precursor material was made of [3]. An oxygen content of 1.1 wt.% was found for the AA6061 powder used and some regions rich in oxides can still be found in fully expanded foams, predominantly near the grain boundaries [5]. This may have an influence on the metallurgical processes during SHT.

(ii) The thermal history of a foamed body is different from that of a conventional Al part prior to heat treatment, giving rise to a different microstructure: alloys intended for heat treatment are usually subjected to an annealing treatment to homogenise the distribution of alloying elements and to dissolve or round off grain boundary precipitates of Mg_2Si or AlFeSi. Aluminium foams cannot be treated accordingly, as heating to the temperatures needed for this purpose - usually higher than in solution heat treatments, namely up to 575°C for alloys of the type examined here [16,17] - might lead to secondary expansion effects [18]. Therefore the post-foaming microstructure resembles more the solidification structure found in cast aluminium with phases rich in alloying elements at grain boundaries.

(iii) The thermal conductivity of aluminium foams is normally more than one order of magnitude lower than that of bulk Al: about 10 $Wm^{-1}K^{-1}$ in a foam with 0.4 g/cm^3 density as compared to about 150-200 $Wm^{-1}K^{-1}$ for dense alloys [7]. Removal of heat during quenching is therefore expected to be much slower.

(iv) Foams are by definition quite heterogeneous structures with struts and membranes of different shapes and cross sections. Quenching is therefore expected to yield different cooling rates for different parts of a sample.

(v) Water quenching of foams could be problematic. Firstly, it would induce high thermal stresses changing the structure of the foam, e.g. by creating cracks. Secondly, as the outer surface of a metal foam as well as the inner cell walls are not free of cracks from the beginning, water can enter the foam after quenching. The pressure drop in gas-filled cells by a factor of almost 3 during transition from 530°C to 20°C will exacerbate this effect. Intruding and evaporating water could have unpredictable effects both on foam structure and on quenching rates.

(vi) The choice of solution heat treatment parameters is limited due to the possibility of unwanted secondary expansion effects.

All these possible problems have to be kept in mind when developing heat treatment procedures for metallic foams.

3 Sample preparation and characterisation

3.1 Preparation of aluminium foams

3.1.1 Selection and preparation of starting materials

Pre-alloyed AA6061 powder (<160 μm) purchased from Mepura (Ranshofen, Austria) was used for preparing the precursor material. The alloy powder was mixed with 0.5 wt.-% titanium hydride powder (Chemetall, Hanau [19]) which acted as a blowing agent. The powder mix was first consolidated by cold isostatic pressing which was followed by hot extrusion [4,5] to long rectangular rods.

Chemical analysis of the extruded material using Atomic Emission Spectroscopy (AES) showed that the alloy composition conforms to the specification for alloy AA6061, except for the Cr- and Ti-content (see Table 2). The latter is a consequence of the addition of 0.5 wt.-% titanium hydride. The age-hardening effect in 6XXX series alloys is primarily based on the occurrence of the Mg_2Si phase in which the mass ratio Mg:Si is 1.73:1. The 0.85 wt.% Mg found in the material therefore correspond to 0.49 wt.% Si in the Mg_2Si phase and only 0.09 wt.-%. excess silicon. Chromium additions are generally believed to be beneficial for reducing grain boundary precipitation. They influence ductility but usually not strength, most significantly in alloys containing excess silicon [20]. Therefore, only limited effects of chromium may be expected for the composition examined here. Other phenomena usually associated with excess silicon can also be ignored.

3.1.2 Foaming

For producing foam specimens a piece of foamable precursor material was inserted into a cylindrical steel mould. Foaming was carried out in a batch furnace with indirect conductive heating and atmosphere circulation at 730°C. During foaming the mould stood upright as shown in Fig. 1 to reduce the influence of transverse drainage. To stabilise the porous structure after foaming two cooling methods were examined: (i) quenching by putting the entire mould into cold water and, (ii) cooling with pressurised air. We shall refer to this quenching step as *F-quenching* in the remaining paper to distinguish it from the quenching step involved in solution heat treatment.

After removal from the mould cylindrical foam specimens were obtained with 44.2 mm diameter, 60 mm height, and an overall foam density of 0.6 ± 0.03 g/cm^3. The closed outer skins which are formed during foaming were not removed because future components to be made from aluminium foam will also have a comparable skin, affecting both strength and susceptibility to heat treatments. As a consequence of drainage during foaming, specimens always exhibit a slight vertical density gradient.

3.2 Heat treatment of Al foams

The heat treatment procedure for aluminium foams was derived from known parameter sets for conventional AA6061 or similar AlMgSi and AlMgSiCu alloys (see Table 1). After solution heat treatment warm ageing was retarded for 3 days to simulate similar delays common in many industrial processes. Heat treatment was carried out in a furnace of the same type as the one used for foaming. The adherence to the required temperatures was tested by measuring sample temperatures with two thermocouples inserted into two small holes drilled into the specimen.

Quenching of the foamed samples after solution heat treatment was carried out in different ways by varying both the quenching medium (air or water) and the conditions of quenching (see Sec. 4.1.1). This quenching step will be called *SHT-quenching* in this paper do distinguish it from F-quenching of the liquid foam described in Sec. 3.1.2. Of all the possible combinations of heat treatment procedures (including the possibility for omitting stages) nine distinct temperature cycles were considered. Fig. 2 gives an overview of the different combinations and a numbering system which will be used throughout the paper.

3.3 Characterisation programme

3.3.1 Quenching tests

In order to be able to assess quenching rates, temperature vs. time curves were measured during SHT-quenching. Such experiments were performed on bare foam samples and also for foams inside the foaming mould. For the former, 2 holes (1.5 mm diameter, 30 mm deep) were drilled into one of the front sides of the foamed cylinders, one close to the skin, one along the axis of the cylinder. Two inconel-encased Ni-CrNi thermocouples were inserted into the holes and pushed inwards to ensure contact with the rearmost cell wall. The thermo-voltage was recorded with a time resolution below 1 ms. Samples were quenched in air and water under various conditions. For quenching inside moulds two holes were drilled into the mould wall through which thermocouples were inserted into the foam (for details see Sec. 4.1.2). Whenever water was used as a cooling medium, it was attempted to seal the holes for the thermocouples to avoid intrusion of water by using a temperature resistant cement.

3.3.2 Studies of meso- and microstructure

Meso- and microstructure of the various specimens were characterised by cutting out thin slices of the cylinders perpendicular to their axes by electrical discharge machining, embedding the slices into black resin and polishing. For discussing the mesostructure of the samples, images of a sector of the circular cross-section were taken at a magnification of 6.5:1. For studying the microstructure magnifications of 200:1 and 500:1 were chosen and images were taken of a cell wall near the centre of a sample as well as of a region near the skin.

3.3.3 Micro-hardness profiles

Slices of 10 mm thickness were cut off the foamed cylinders for hardness tests and were infiltrated with resin from both (open) face sides, after which one face was polished. The visible open pores were usually completely filled with resin so that the cell walls standing approximately perpendicular to the polished surface were fully supported against buckling. Vickers hardness with a load of 50 g and a loading time of 10 seconds (HV 0.05) was measured at various locations in the sample. The locations of measurement were determined by first defining points in approximately equal steps along an ideal line leading from a sample's centre to its skin. Whenever the point chosen did not lie in the metallic matrix, a point in the centre of the next available foam membrane was used. The diagonal extension of the indentations ranged from 27 to 70 μm. Testing standards [21] require a distance from the centre of the indentation to the sample edge of 3 times this value for light metals like aluminium. In aluminium foams, a cell wall thickness in the region of 80 μm is common for AlSi7 matrix alloys, and similar, though slightly higher values of approximately 110 μm have been found for AlMg1SiCu [22]. Thus the standard's requirements cannot always be satisfied. Therefore we checked all indentations for possible asymmetries which would indicate insufficient support by the resin on the side closest to the next wall boundary. As neither such asymmetries nor any tendency for lower apparent hardness values in thin sections were found, the results are considered sufficiently reliable for the purpose of the present work.

3.3.4 Compression tests

Quasi-static compression tests were carried out on a Zwick 1474 testing machine at 20°C and at a constant crosshead velocity of 5 mm/min. For each heat treatment state 4 different specimens were tested. Compression was stopped whenever either 80% strain or 95 kN force (equivalent to 61.9 MPa) were reached.

4 Results and discussion

4.1 Characterisation of quenching behaviour

4.1.1 SHT-quenching

Quenching was carried out in various ways: first, either air or water quenching was applied. Then, for each quenching medium three different modes were chosen: resting air, mildly pressurised air (comparable to a standard table fan) and pressurised air (6 bar laboratory air supply). Water quenching was carried out, (a) with a partially immersed sample without movement, (b) with a completely immersed sample and as much movement as the attached thermocouples allowed it and, (c) with violent movement, the letters referring to Fig. 3. Except for case (c) the temperature could be recorded during quenching. Fig. 4 shows some of the resulting temperature curves for two quenching tests in water. It is obvious that in both cases the region near the skin is quenched at a much higher rate compared to the centre of the cylinder. For mode “a” a near-exponential decline of temperature – time-delayed for the centre of the sample – can be observed. For mode “b”, however, some non-exponential features can be detected (marked by dotted circles) which can be attributed to the intrusion of water into the hole drilled for the thermocouples in spite of the measures to prevent this. This leads to higher apparent quenching rates as the thermocouple is cooled directly. Such effects were only observed when the sample was completely immersed into the water (“b” in Fig. 3), while they did not occur when the intrusion of water into the holes drilled for the thermocouples was prevented (“a” in Fig. 3).

In Fig. 4 the critical temperature range from 290°C to 400°C is marked. Average and maximum quenching rates in this range can be determined by either calculating differences $(400°C-290°C)/(t_{400}-t_{290})$ or by taking the peak value of dT/dt for said temperature range. Average quenching rates are shown in Fig. 5 for all the different quenching modes applied. As could be expected the quenching rates for the quenching medium water are much higher than for air. Moreover, the difference between centre and skin is much larger, as for air quenching the thermal conductivity of the aluminium foam is high enough to level out major radial temperature gradients. It is also obvious from the SHT-quenching rates displayed in Fig. 5 that only quenching in water yields the quenching rates necessary for achieving an ideal or almost ideal supersaturated solid solution in the alloy investigated (>100 K/s; the sample core falls short of this value). At least at the air flow rates used here, air-quenching is not fast enough for this purpose as the cooling rates achieved did not exceed 10 K/s. In contrast, the quenching rates measured for moderate movement in water are about 36 K/s in the centre of the sample. We estimate that the strongest agitation (which was applied when making the samples for the mechanical tests) leads to about 50 K/s. Therefore, the sample centre could possibly show a non-optimised age-hardening, an effect which will gain importance when even larger cross sections are examined.

4.1.2 F-quenching

The F-quenching temperature curve could not be obtained by measuring the temperature during the actual solidification as this would have led to difficulties with holding the thermocouples in place in the semi-liquid expanding foam. Instead, pre-manufactured cylinders were placed inside the mould, fitted with thermocouples, heated to 600°C and were then quenched. The results obtained this way should yield data close to the real situation in F-quenching. Not accounted for in such measurements are the slightly higher temperatures during foaming and the latent heat of the solidifying metal. Both F-quenching in air and water (results not shown) yield cooling rates which are significantly lower than the ones observed during SHT-quenching. This is readily understood, as in this case it is not only the foamed part itself that has to be cooled but also the mould in which it was produced. Consequently, the amount of heat to be dissipated is larger, while the various additional interfaces (foam → air gap → steel) reduce conduction. It is evident that such results strongly depend on the material and design of the mould so that they merely represent a very special situation. Nevertheless, with F-quenching rates using water well above 10 K/s it is considered worth trying to age water F-quenched parts without a prior SH-treatment (#2). The same procedure is applied to a number of samples that were F-quenched using air (#6).

4.2 Meso- and microstructure

One can distinguish between mesoscopic – the millimetre range – and microscopic – the micrometre range – length scales in foamed metals. Mesostructure is governed by the flow of metal during foaming and processes during subsequent F-quenching. It will influence the overall mechanical behaviour of the foamed structure. Microstructure is first created during F-quenching but is finally adjusted only during heat treatment of the solid foam. Microstructure determines the mechanical properties of the building elements of the mesostructure, namely the cell edges and walls.

Meso- and microstructure will be discussed for four different samples: air and water F-quenched foams and, for the former, foams without subsequent heat treatment, fully heat treated samples and an annealed foam. Figs. 6-9 show the corresponding images. The investigation is limited to light microscopy. The careful TEM analysis necessary for resolving the structure of the precipitates responsible for age-hardening is beyond the scope of the present paper.

4.2.1 Mesostructure and foam morphology

A comparison of the low magnification images in Fig. 6 shows that the type of F-quenching has a considerable effect on larger scale structural features: first of all, the air-quenched foam (# 4) has a lower density and correspondingly larger cells and thinner cell walls near the centre, whereas the water-quenched sample (#1) shows comparatively small cells and thicker cell walls. The explanation for this is that air cooling leads to a slower solidification of the foam starting from the outer skin, giving rise to a situation in which the central part still expands while the skin has already solidified. Therefore, material is redistributed in radial direction during cooling.

On the other hand, water quenching solidifies the entire structure in a much shorter time. Looking at the higher magnification images Fig. 7) a further difference becomes evident: the foam cooled in water after foaming (#1) is characterised by smoother cell walls and fewer cracks within these walls, whereas the air-quenched sample (#4) has more wavy cell walls and more cracks. Cracks may be created in the stage in which the outer skin is already solid but the foaming core still expands, thus creating stresses in regions already solidified.

This tendency becomes even more significant after a precipitation hardening treatment (Fig. 8). It is obvious that SHT-quenching widens existing cracks or creates new ones due to the thermal stresses during the treatment and therefore has a detrimental effect on foam morphology and most probably on mechanical properties.

4.2.2 Microstructure and distribution of precipitates

The microstructures shown in Fig. 9 confirm the picture drawn above. The samples F-quenched in air without any further heat treatment (#4) is dominated by grain boundary precipitates in which, by means of EDX-analysis, Mg_2Si- and AlFeMnSi-phases could be identified. The concentration of precipitates is strongest in the sample centre where the quenching rate was lowest.

Precipitation hardening of such samples alters this situation: the fully heat treated sample (#8) has a lower precipitate density while grain boundaries are still visible, though slightly less dominant and less localised. The phases found in the cloudy grey areas are similar to the ones detected in the initial state, suggesting that a complete solution of Mg_2Si-phases was not achieved during SHT. However, as the concentration of these precipitates is lower compared to the untreated samples, it can be presumed that a part of the alloying elements must have formed strength-increasing GP-zones.

In contrast to this, annealing foams results in a completely different microstructure (#5): Mg_2Si equilibrium precipitates have increased in size and volume fraction. They are once again mainly located at or near the grain boundaries and often form a pattern of needle-shaped inclusions. The grains themselves are for the most part devoid of alloying elements which explains the low strength of this metallurgical state (see Sec. 4.3 and 4.4).

4.3 Micro-hardness

Micro-hardness profiles were obtained for specimens in all different heat treatment states by applying the procedure described in Sec. 3.3.3. Between 26 and 41 points were chosen for each profile. Hardness was usually measured in vertices or in sufficiently thick cell walls, trying to avoid areas where too thin or porous cell walls or hidden pores could give rise to artefacts.

Fig. 10 shows hardness profiles for three different heat treatments. For these and for all other samples examined, no significant functional relationship between the hardness value and the radial position of the measurement could be established, i.e. no general tendency towards lower hardness in the central region became evident as might have been expected for insufficiently age-hardened samples. This shows that despite the low thermal conductivity and its local variation a fairly uniform age-hardening effect can be achieved. The fluctuations of hardness around an average value could have four possible reasons: i) heterogeneities in the cell wall due to the specific nature of the material containing oxide inclusions, micro-pores etc., ii) possible artefacts originating in the difficulty of measuring the hardness of thin structures due to a possible elastic rebound (although nodes and thick cell walls were selected for this reason), iii) microstructural differences between areas within a grain caused by incomplete homogenisation after foaming and, iv) different quenching rates due to non-uniform thermal conductivities in the heterogeneous foam. The relative importance of these effects can be estimated by comparing different heat treatments. As annealing does not include any quenching the observed low scatter in hardness shows that the first three mechanisms must be quite small. In contrast, the untreated sample (#4) and even more the age-hardened sample (#8) exhibit much stronger fluctuations, showing that a great part of them has to be attributed to local differences in thermal history or to the assumed concentration gradients, i.e. mechanism iv) is most important.

The obviously statistical nature of hardness variations suggests calculating an average of this quantity over the entire sample thickness. Fig. 11, right axis, displays such averages for all the heat treatments which were considered. The effects expected for the different heat treatments can indeed be identified: Annealed specimens (#5) exhibit the lowest values, followed by those without heat treatment ("as foamed", #1 and #4), the two F-quenching methods "air" and "water" yielding about the same results. The specimens that were warm-aged directly after foaming without a prior SH-treatment (#2 and #6) show an increased hardness level. The difference between air- and water-cooled samples is well within the range of the experimental error. F-quenching by itself therefore already leads to some supersaturation of the alloy which allows for the formation of precipitates. The highest hardness values were measured for the solution heat treated and warm aged specimens (#3 and #8). SHT-quenching is therefore more effective in creating a supersaturated alloy than F-quenching with its lower quenching rate. The values measured for the water quenched specimens are higher than those for the air-quenched ones, possibly reflecting the higher quenching rates which can be achieved with water. However, one should keep in mind the quite pronounced scatter of the measurements when drawing such conclusions (see error bars). Sole SH treatment without subsequent ageing (#7) yields hardness values that are comparable to those measured for warm-aged ones without prior solution heat treatment (#2 and #6). Thus they are substantially higher than those for samples without heat treatment.

4.4 Compression behaviour

4.4.1 Stress-strain curves

Fig. 12 shows stress-strain curves for the different specimens tested. Fig12(a) displays test results corresponding to individual samples, whereas for a more detailed comparison of the various heat treatments groups of four curves measured on equivalent samples have been averaged in Fig 12(b) and (c).

Two different types of curves can be distinguished. The first type is typical for ductile alloy foams. These fail with smooth and steadily rising stresses. In contrast, brittle foams show a stress peak if strained beyond the elastic regime and a corresponding drop in stress afterwards [23-25]. The various curves measured for foams in different states of heat treatment can be grouped into these two categories. Processing routes #1, 4, and 5 show the typical behaviour of ductile alloys. Processing routes #3 and #8, i.e. both precipitation hardened alloys, show a compressive response typical for brittle alloys. The remaining samples show aspects of both types: sample types #2 and #6 show a predominantly brittle behaviour, sample #7 is closer to the ductile limit. The different types of stress-strain-curves stem from differences in their failure modes [26]. In samples showing smooth and constantly rising stresses, failure is controlled by bending rather than breaking of cell walls and struts. As strength is density-dependent, a specimen with slightly inhomogeneous density distribution will first be deformed in the section with lowest density (“weakest link”) [24]. With this region being simultaneously compressed to higher densities and strain-hardened, a new deformation band will soon develop and take its place [27]. In contrast to this, brittle failure does not necessarily start at a weakest link. Cracks created by thermal stresses induced during production of the foam or heat treatment may become starting points for failure of the structure as a whole. Strain-hardening effects will not occur. Furthermore, failure characterised by breaking off of specimen parts may result in sections that represent only part of a sample’s theoretical cross section. The local stress within these sections is increased, while the integral stress values calculated on the basis of a theoretical cross section may even decrease. Such a decrease is clearly visible for specimen #8 and can still be distinguished for specimen #6 - both were F-quenched using air. Their counterparts which were F-quenched in water, particularly #3 and #2, show some deviation from this typical response: An initial peak with a subsequent, immediate drop in stress levels is distinguishable, but in both cases, it is not followed by a drop in stress levels over a large strain range as seen typically for specimen #8. Instead, throughout the strain range in question, a significant rise in stress levels is observed.

Fig. 12(a) quite clearly shows that the scatter between the various individual but equivalent samples is lowest for the most ductile, the annealed specimen (#5). Foams in the state “as foamed” (#4) and even more the foams with the highest strengths (#8) show a much higher scatter in their stress-strain response. This is a clear analogy to hardness. A possible reason for this difference is that ductile failure by bending of cell elements is a process governed mainly by the local strength of the cell material, whereas brittle behaviour depends also on features like local cell shapes and crack arrangements. Clearly, the former property is more reproducible than arbitrary morphological features. However, cell wall strength might also be non-constant over the extent of the sample owing to varying cooling conditions during F- and SHT-quenching. Such varying conditions could be caused by material inhomogeneities which lead to different quenching rates in different areas of the foam. As annealing does not involve any quenching, the high reproducibility for theses samples is readily understood.

4.4.2 Compression strength

Definition of compression strength is somewhat ambiguous for metal foams. As one observes a different stress-strain behaviour depending on the type of alloy used, varying definitions for strength can be adequate [6]. For determining strength values from the stress-strain curves measured in this work the following approach was chosen: strength is taken as the initial peak stress wherever it occurs in Fig. 12(b) and Fig. 12(c), i.e. for heat treatments #2, 3, 6 and 8. For samples that exhibit no peak stress, i.e. for #1, 4, 5 and #7 stress values obtained at 5% total deformation were used. This strength is compared to hardness in Fig. 11. The two scales have been adjusted such that the two quantities appear identical for sample #4. One immediately realises that strength and hardness correlate very well.

Specimens that went through a complete precipitation hardening cycle assume the highest strength values which are 75% higher as compared with the untreated foam for F-quenching in air (#8) and 60% higher for F-quenching in water (#3). These levels of strength increase are higher than known from work previous work [6-9]. Foams that were simply aged without SH-treatment (#2 and #6) show lower, but still rather high strength levels which are 40 to 45% higher than strengths of untreated samples. Obviously, F-quenching after foaming alone leads to a degree of supersaturation – slightly more for the faster F-quenching in water than for F-quenching in air – which is sufficient for the following ageing to produce precipitates suitable for increasing strength considerably. However, F-quenching is too slow to generate a perfect supersaturation and therefore full SH-treatment leads to even better results.

Foams which were only SH-treated (#7) have about the same strength as the “as foamed“ samples, which is a bit contradictory to the picture presented by the hardness measurements, where an increase was observed after this treatment. Annealed specimens (#5), however, show a strength which is lowered by 35% as compared with untreated reference samples. The reduction of matrix strength associated with grain growth and formation of Mg_2Si along the grain boundaries is the reason for this behaviour.

All strength values measured exhibit a much lower statistical scatter than hardness. This is because hardness is a very local quantity, whereas strength reflects a property averaged over many cells. This is exacerbated by the incomplete microstructural homogenisation which created local fluctuations and may account for considerable differences in local hardness within a single cell wall.

4.4.3 Energy absorption

Energy absorption is an important technological property of foams. One would like a rectangular stress-strain curve which, however, cannot be achieved in real materials. From the stress-strain curves energy absorption efficiencies were calculated using the definition:

$$\eta(\varepsilon) = \frac{\int_0^{\varepsilon} \sigma(\varepsilon') \cdot d\varepsilon'}{\max \sigma(\varepsilon)\big|_0^{\varepsilon} \cdot \varepsilon} .$$

In Fig. 11 the efficiency for a given compression level of 50%, $\eta(50\%)$, and the maximum efficiency over the entire compression cycle, $\eta_{max} = \max \eta \big|_{0\%}^{80\%}$, are shown. The stress-strain-curves observed for ductile matrix materials are not completely satisfactory when high efficiency of energy absorption is required. Highest efficiency is obtained when the plateau region is horizontal and no peaks occur. Thus the most promising energy absorption behaviour with high efficiency over an extended strain range must be attributed to specimens of processing route #6. For all samples treated in this way a non-typical failure mode combining a peak stress known from brittle failure with a constant rise of stress levels known from ductile foams was observed. Therefore age hardening without prior solution heat treatment in combination with F-quenching using air seems to result in increased strength and a reduced but not completely eliminated ductility. This route could therefore represent a technologically relevant compromise between the various requirements in constructing an efficient energy absorber.

5 Conclusions and outlook

The experiments have shown that heat treatment is a viable option to customise the properties of aluminium foams. For specimens based on the wrought aluminium alloy AA6061 with a density of 0.6 g/cm^3, strength could be varied between 7 and 20.5 MPa. The results apply to the sample geometry selected for the present study but as in practice foamed components will in many cases not be much thicker than the cylinders used, the conclusions of this paper can be readily extended to other geometries provided that these are not too bulky, such as e.g. flat panels or any kind of small component. Full heat treatment cycles yielded the highest strengths.

A possible drawback of using water for quenching is the intrusion of water into the foam structure during quenching. However, F-quenching in water seems to lead to a better meso-structure of foams, which is highly desirable. Therefore, the initially sceptical view towards water quenching had to be partially revised, and an adaptation of mould concepts taking possible benefits of this process route into account may well be worth considering.

In this context it is encouraging that ageing directly after foaming without SH-treatment also yields a considerable yet slightly reduced age-hardening effect. Another positive effect is that specimens treated in this way show a higher energy absorption efficiency and also a smaller statistical scatter than samples subjected to a complete precipitation hardening cycle. Therefore, especially for applications in energy absorption F-quenching using air may be a better choice. The phenomenon itself deserves continued investigation.

Future developments will include a further adaption of heat treatment parameters to specific alloys, e.g. other wrought 6XXX alloys or even high strength 7XXX alloys.

Among these, special interest should be directed towards alloys with a greater tolerance regarding quench rates - mostly alloys with a reduced copper content, or even without copper as an alloying element, such as AA6082 as a replacement for AA6061 or AA7020 instead of AA7075. Moreover, additives could be sought which refine the initial solidification microstructure of the alloys after foaming and compensate the missing homogenisation treatment.

References

[1] M. F. Ashby, A.G. Evans, N.A. Fleck, L.J. Gibson, J.W. Hutchinson, H.N.G. Wadley, Metal foams – a design guide, Butterworth-Heinemann, Boston, 2000.

[2] J. Banhart, Prog. Mater. Sci. 46 (2001) 559-632.

[3] J. Banhart, Journal of Metals 52 (2000) 22-27.

[4] F. Baumgärtner, I. Duarte, J. Banhart, Adv. Eng. Mater. 2 (2000) 168-174.

[5] I. Duarte, J. Banhart, Acta Mater. 48 (2000) 2349-2362.

[6] J. Banhart, J. Baumeister, J. Mater. Sci. 33 (1998) 1431-1440.

[7] F. Simančík, J. Jerz, J. Kovačík, P. Minár, Kovové Materiály/Metallic Materials 35 (1997) 265-277.

[8] F. Han, Z. Z hu, J. Gao, Met. Trans. 29A (1998) 2497-2502.

[9] R. Gradinger, F. Simančík, H.P. Degischer, in: S. Felber, T. Varga, J.L. Zeman (Eds), Welding Technology, Materials and Materials Testing, Fracture Mechanics and Quality Management, Vol. 2, 1997, Technical University of Vienna, Vienna, pp. 701-712.

[10] J.R. Davis (Editor), Aluminum and Aluminum Alloys, ASM International, Materials Park, 1993.

[11] M. F. Ashby, D.R.H. Jones, Engineering Materials, Pergamon Press, Oxford, 1980.

[12] T. Lyman (Editor), Metals Handbook, 8th Edition, ASM, Metals Park, 1964.

[13] H. Bomas, Die Abschreckempfindlichkeit manganhaltiger AlMgSi-Knetlegierungen, Fortschrittsberichte der VDI-Zeitschriften, VDI-Verlag, Düsseldorf, 1980.

[14] A. Latovski, M. Bronicki, Aluminium 65 (1989) 391-397.

[15] H. Suzuki, M. Kanno, G. Itoh, Aluminium 57 (1981) 628-629.

[16] E. Cyrener, W. Härtel, Neue Hütte 29 (1984) 420-426.

[17] E. Cyrener, M. Krumnacker, Neue Hütte 30 (1985) 130-136.

[18] M. Hahn, A. Otto, Mat.-wiss. u. Werkstofftech. 31 (2000) 432-435.

[19] R. Kresse, in: J. Banhart, M.F. Ashby, N.A. Fleck (Eds), Metal Foams and Porous Metal Structures, MIT-Verlag, Bremen, 1999, pp. 109-112.

[20] R.C. Dorward, C. Bouvier, Mater. Sci. Engnr. A254 (1998) 33-44.

[21] Testing Standard DIN EN ISO 6507-1, Beuth Verlag, Berlin, 1997.

[22] H. Stanzick, M. Wichmann, J. Weise, L. Helfen, T. Baumbach, J. Banhart, Advanced Engineering Materials, in press (Oct. 2002)

[23] B. Kriszt, B. Fouroughi, A. Kottar, H.P. Degischer, in: T.W. Clyne and F. Simančík (Eds), Metal Matrix Composites and Metallic Foams, Wiley-VCH, Weinheim, 2000, pp. 74-82.

[24] F. Simančík, in: J. Banhart, M.F. Ashby, N.A. Fleck (Eds), Metal Foams and Porous Metal Structures, MIT-Verlag, Bremen, 1999, pp. 235-240.

[25] P.H. Thornton, C.L. Magee, Met.Trans. 6A (1975) 1253-1263.

[26] L.J. Gibson, M.F. Ashby, Cellular Solids: Structure and Properties, 2nd Edition, Cambridge University Press, Cambridge, 1997.

[27] A.-F. Bastawros, A.G. Evans, in: J. Banhart, M.F. Ashby, N.A. Fleck (Eds), Metal Foams and Porous Metal Structures, MIT-Verlag, Bremen, 1999, pp. 221-226.

Figure Captions

Fig.1. Precursor material, mould, aluminium foam cylinder and cross section through foam (from left to right).

Fig.2. Overview over the different heat treatment procedures investigated in the study. Numbers are used in the text to specify heat treatment as ‘#n’.

Fig.3. Three ways to carry out SHT-quenching experiments in water: (a) partial immersion without moving, (b) complete immersion while moving moderately, (c) violent movement.

Fig.4. Solution heat treatment (SHT)-quenching curves carried out in mode “a” and “b” as defined in Fig.3. Temperatures near the centres and the skins of the samples are given. For mode “a” an exponential fit curve is displayed (dotted line). The dotted circles indicate strong deviations from exponential behaviour for mode “b”. The critical temperature range for premature formation of precipitates ($290°C<T<400°C$) is marked by horizontal lines.

Fig.5. Measured SHT-quenching rates for various quenching modes. Average rate in critical interval $290°C < T < 400°C$ is given.

Fig.6. Cell morphology of two AA6061 specimens in different states of heat treatment. Left: F-quenched in air (#4), right: F-quenched in water (#1). Both samples without any further heat treatment.

Fig.7. Magnification of cell features in the foams shown in Fig. 6 (top: #4, bottom: #1)

Fig. 8. Cell morphology of two AA6061 foams both F-quenched in air. Top: no further heat treatment (#4), bottom: full SHT (#8). Circles mark cracks created by SHT.

Fig. 9. Microstructure of three AA6061 foams, all F-quenched in air. Top: no further heat treatment (#4), middle: full SHT (#8), bottom: annealed sample (#5)

Fig. 10. Micro-hardness profiles for specimens with different heat treatments (F-quenched in air). Full line shows averaged trend.

Fig. 11. Right axis: spatially averaged micro-hardnesses for all the different heat treatments. The error bars are a measure for the spatial fluctuations seen in Fig. 10. Left axis and upper panel: compressive strength and energy absorption efficiency (values were averaged over 4 individual measurements).

Fig.12. Stress-strain-response of heat treated aluminium foams. a) four individual measurements are shown for three different heat treatments; b) and c) averaged stress-strain curves are shown for 8 different heat treatments, where b) F-quenched in air, c) F-quenched in water.

Tables

Table1. Review of parameters for age-hardening and annealing of AA6061 alloys. Parameters in bold were used in the present studies for heat treating foams.

Solution heat treatment				**Warm ageing**		**Comment/Source**
T_s [°C]	t_s [min]	$(dT/dt)_q$ [K/s]	t_d [min]	T_a [°C]	t_a [h]	
530	-	100	-	160	-	[10]
530	-	100	-	170	-	[10]
530	-	100	-	175	-	[10]
540	40	-	80	160	-	[13]
525	25	-	-	180	-	[14]
560	30	-	-	100	166,7	[15]
				175	16,7	two-step ageing process [15]
530	**100**	**-**	**3 days**	**165**	**10**	**SHT air-quenched** [this work]
530	**100**	**-**	**3 days**	**165**	**10**	**SHT water-quenched** [this work]
Annealing						
$(dT/dt)_h$ [K/min]	$T_{an,1}$ [°C]	$t_{an,1}$ [h]	$(dT/dt)_{an,1\to2}$ [K/h]	$T_{an,2}$ [°C]	$t_{an,2}$ [h]	
13.8	**415**	**1**	**15**	**230**	**2**	[this work]

Table 2. Chemical analysis of foamable precursor material (AA6061 + 0.5wt.% TiH_2) used for foam production compared to specifications of alloy AA6061 [10].

element		Mg	Si	Cu	Fe	Mn	Cr	Ni	Zn	V	Ti
measured	[wt.%]	0.85	0.58	0.20	0.18	0.002	0.001	0.001	0.011	0.017	0.69
reference	min.→	0.8	0.4	0.15			0.04				
values for	[wt.%]	-	-	-	0.7	0.15	-	-	0.25	-	0.15
AA6061	max. →	1.2	0.8	0.40			0.35				