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Processing, stabilization and applications of metallic foams. Art of science

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Metallic foams have been of considerable interest both from the industrial applicability and scientific viewpoint. In the last three decades, since the inception of the metal foam, several processes have been developed to introduce large size pores, almost uniformly distributed, in the metallic materials. These processes include liquid metal route, powdered metal technique and some other methodologies in which pores are incorporated by direct sintering of foamed elements. Metallic foams have seen tremendous innovations for the last few years and efforts are being made to achieve foams with a low cost and with reliable properties. The stability of cell structure during the foaming process has also been a concern for researchers. Therefore, methods have been explored to increase the stability by introduction of ceramic particles in the material to be foamed, addition of alloying elements, secondary processing of precursor materials, and by optimizing the process parameters. To this end, experiments have also been made in microgravity conditions to study the mechanism of foam formation which can give a better understanding for controlling the processing conditions. Despite the current development, there are several limitations for a wide applicability of metallic foams in various industrial sectors, such as high cost of production, variability in measured properties, and unavailability of reliable testing methods. Taking these facts into consideration, the present paper is aimed at reviewing various foam processing techniques, the measures required to be taken for making stable foams and the industrial applications of the metallic foams in the current state of development.

1. Introduction

The stringent requirements of the market and the economics associated with light materials have driven our efforts to develop metallic foams with porosity ranging between 40 and 98 vol. %. In last three decades, several processing techniques have been developed to manufacture metallic foams [1–12]. The basic objective behind this developmental process is the unique combination of physical and mechanical properties expected from metallic foams such as high stiffness, low specific weight, high gas permeability, low thermal conductivity, unusual acoustic properties, high impact ab-

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sorption capacity, and good electrical insulating properties. The potential applications of metal foams have been envisaged in the area of automotive industries, light weight construction materials, silencers, flame arrester, heaters and heat exchangers, catalysts, electrochemical applications, military armour vehicles and aircraft industries [1, 12–15].

In general, a foamed material is understood to have a large volume fraction of gas pores uniformly dispersed in a liquid. A single gas inclusion is separated by liquid film making a cell. This morphology, however, can be preserved during solidification of liquid leading to solid foam. When we talk about metal foam, it is understood to have uniformly distributed gas pores with volume fraction in the range 40–98%. This kind of morphology is also known as cellular solid structure. There are three variants of cellular solids: open cell, closed cell structures [1], and a combination of the two [8]. Recently, a novel structure has been developed, which is known as lotus-type growth [4, 16]. The open cell structures incorporate interconnected pores, whereas close cell pores are surrounded by a metallic wall. The lotus-type morphology consists of long cylindrical pores aligned in one direction. These foam structures are shown in Fig. 1. However, in the subsequent sections we will be dealing mainly with closed cell and open cell structures.

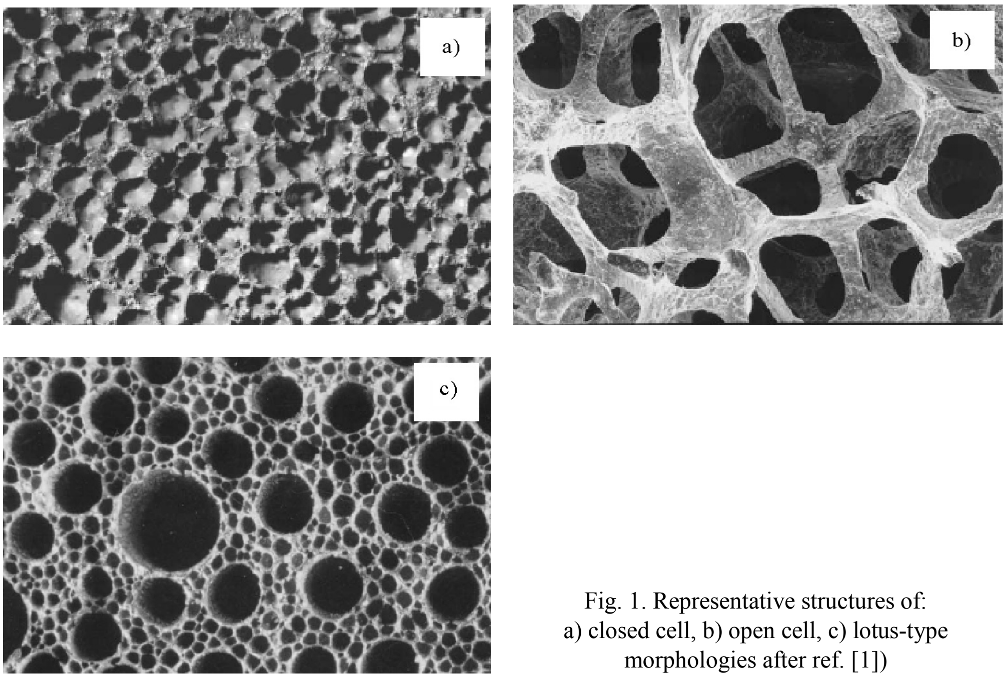

Fig. 1. Representative structures of: a) closed cell, b) open cell, c) lotus-type morphologies after ref. [1])

Depending on the requirements, several metals and alloy foams have been developed in the recent years, e.g. Al, Al–Si, Al–Mg [1], Cu [4, 6, 11, 16], Pb [17], Fe, steels [18–21], Ni_3Al [22], Zn [23], Mg and Ti [24], Al–Cu, MMCs etc. [1]. Out of these foams, tremendous development has taken place in the field of Al foams with some industrial applications. However, a number of research groups throughout the world have been involved in the optimization of processing routes and, thereby, reduction in the cost of production of metal foams [25]. In general, two processes have been invariably used for metal foaming: (1) the liquid metal route where foaming is accomplished by direct foaming of melt with gas or some foaming agent [1, 26–29] and (2) the powder metallurgy route where foaming is affected by foaming a sintered compact [1]. Other routes include sintering of hollow spheres [30] and pressing the material around a filler followed by sintering with subsequent filler dissolution or decomposition [9–11, 31, 32]. Recently, experiments have been carried out in microgravity conditions to see its effect on the stability of foams and obtaining highly porous structures [33–36].

In view of the above facts, the aim of this paper is to review the recent developments in metallic foams. We will discuss various production techniques along with their advantages and limitations, the effect of processing parameters on the stability and/or morphology of foam structures, industrial viability of these techniques, and the potential of metallic foams as a prime structural material in the next generation.

2. Processing routes for metallic foams

There are many ways by which cellular metallic materials can be produced. As aqueous or polymeric liquid foaming is well established, we will consider only the processing of metallic materials. A processing map for the same is shown in Fig. 2.

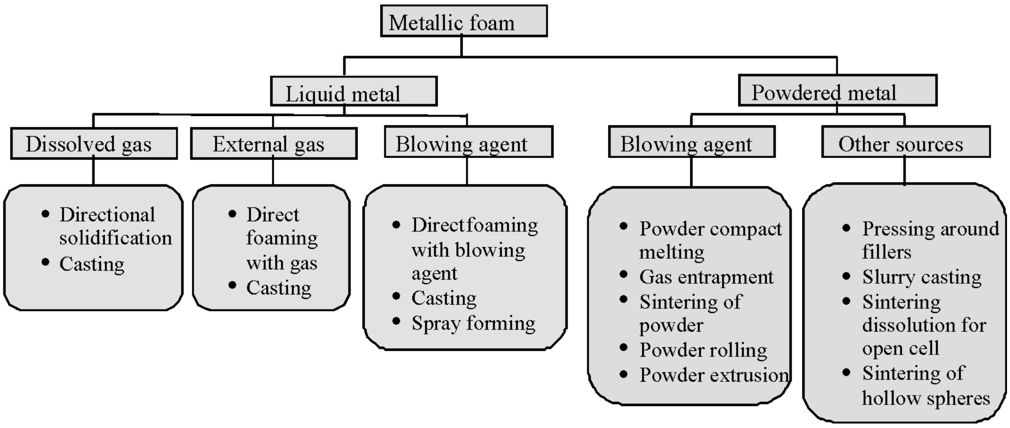

Fig. 2. The processing tree for metallic foams

This indicates the possibility to produce metallic foams in two materials states: liquid metal and powdered metal. As we know that the basic aim of foam processing is to incorporate large and uniformly distributed gas pores in the metallic materials, we have summarized it as the gas source used in the two states of matter mentioned above. The external gas source means the melt has to be foamed with the help of gases such as air, nitrogen or argon. Dissolved gases lead to foaming because they precipitate from the melt during solidification and if they are not allowed to escape, they will give rise to a foamed structure. The gas source from a blowing agent implies decomposition of a chemical giving rise to a gaseous product. The gas generated from this source is entrapped in liquid or semi-solid material culminating into metallic foam. In the following sections, we will review various processes where different gas sources or other materials are utilized to produce cellular materials.

2.1. Liquid state processing

The liquid state foaming involves three different gas sources: external gas source, gases generated by some foaming agent and dissolved gases. The external gas source to make aluminium foam was first patented by Alcan International and utilized by Hydro Aluminium (Norway) and Cymat Aluminium (Canada) for commercial production [37]. This technique has also been used for the production of foams from metal matrix composites.

External gas source. This process consists in the blowing of liquid metal by gases and to ascertain uniform distribution of large gas pores in the liquid metal. This is accomplished by creating very fine gas bubbles in the melt via especially designed rotating impellers as shown in Fig. 3. However, the stability of the foamed structure is an important factor. The movement of gas bubbles in the liquid becomes relatively easy when the liquid melt viscosity is very low, and the bubbles segregate and coalesce together giving rise to large pores. Keeping this in view, melt viscosity is enhanced by incorporating some second phase ceramic particles, e.g. SiC, Al_2O_3 or MgO [1, 39–41]. As the incorporation of ceramic particles in the liquid metal depends on

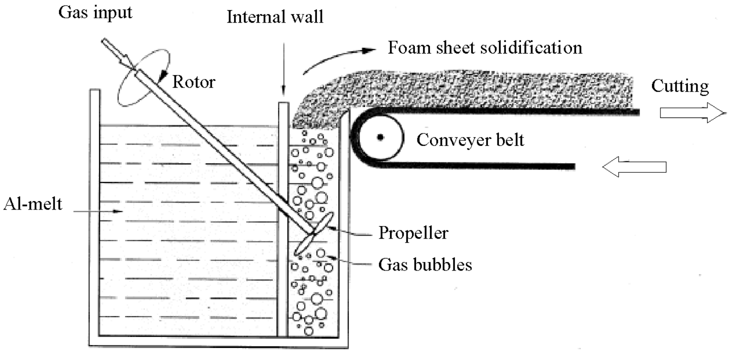

Fig. 3. A scheme showing direct foaming of liquid melt by gas injection [38]

several factors, the first step in this kind of processing is to make metal matrix composite followed by gas injection. The stabilization of the foamed structure obtained via this route is derived from the presence of particles in the cell wall. Due to a high contact angles of most of the ceramic particles with Al melt, there are chances of particle strip off from the cell boundaries leaving the wall weak. Therefore, the imperative should be to select particles having good wetting with the liquid metal so that particles are not stripped off the cell wall. A very low contact angle also does not ensure the stabilization effect [1]. The volume fraction of the particles may range from 10% to 20% with a mean particle size of 5–10 μm. However, the size and fraction of particles for a good foaming have been empirically established as shown in Fig. 4 [39, 42].

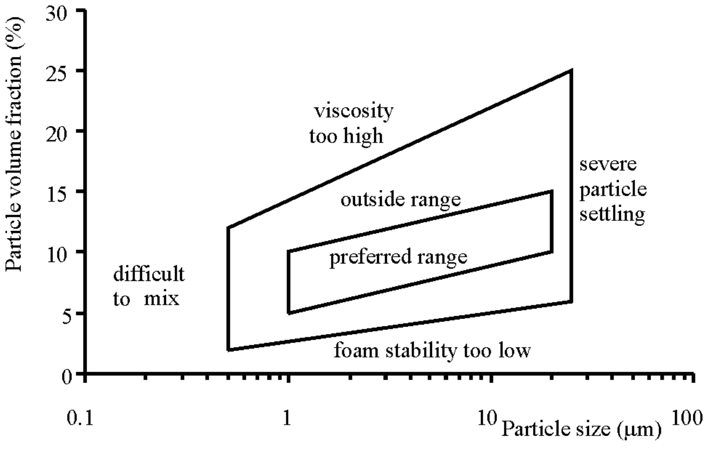

Fig. 4. Empirical guide for the selection of particle size and its content for foaming MMC melts (after ref. [42])

Particles smaller than 1 μm are difficult to mix, and on the other hand particles bigger than 20 μm lead to severe particle settling. A low volume fraction does not stabilize the foam, whereas a high fraction of particles culminates into high viscosity leading to a difficulty in gas injection. This technique has been extensively used for foaming Al and its alloys, and porosities in the range of 80–95% with 3–25 mm cell size have been achieved. The process is very effective in producing large size foams continuously; however, the presence of ceramic particles leads to difficulties in machining due to high hardness and brittle behaviour of foams. To avoid these disadvantages, it was suggested that foaming should be carried out in a melt, without second phase particles, very close to its liquidus temperature with an arrangement to continuously cool the liquid metal during bubbling. This helps in keeping the viscosity of the melt at low level.

Blowing agent as gas source, direct foaming. As an alternative to the direct foaming of melt by an external gas source, a blowing agent can be added to a viscous melt, which decomposes on heating and releases gas leading to the foaming process. This technique is being commercially exploited by Shinko Wire (Japan). Also in this processing route, the enhancement of melt viscosity is a prime requirement which can be accomplished by the incorporation of calcium, ceramic reinforcements, aluminium dross, metallic viscosity-enhancing additives and manganese dioxide, etc. [1, 39, 43]. A scheme of the process is shown in Fig. 5.

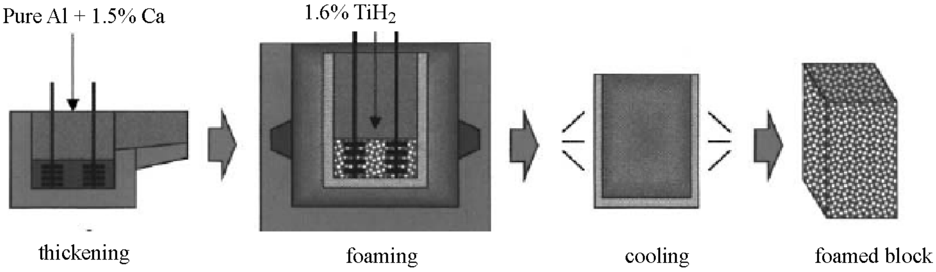

Fig. 5. Processing steps involved in liquid melt foaming with addition of a blowing agent [26]

For example, the addition of Ca to Al melt forms oxides and intermetallics (CaO, $CaAl_2O_4$ and Al_4Ca) which increase the viscosity by thickening the melt. To achieve high viscosity of the melt ($(6–7)\times10^{-3}$ Pa·s), a continuous stirring for up to 8–12 min is required for calcium content of 1.8 wt. % [44]. The process is carried out by melting Al in a crucible followed by addition of about 1.5–3.0 wt. % Ca. Subsequently, the melt is stirred and a blowing agent (namely, 1.6 wt. % TiH_2 for Al) is added. The blowing agent releases hydrogen gas under the influence of hot viscous liquid, leading to gradual foaming of the melt. A careful control of the process parameters such as uniformity of temperature of the melt, viscosity and distribution of TiH_2 particles give rise to a homogeneous foam structure. The foaming is generally carried out at 680–720 ºC. Instead of TiH_2, ZrH_2 (0.5–0.6 wt. %) can also be used for the production of Al foam with a preferable temperature between 670 ºC and 700 ºC. Typical densities of such Al foams have been determined to be around 0.18–0.24 $g\cdot cm^{-3}$. In some cases, it becomes difficult to mix the blowing agent in the liquid metal as it decomposes very fast at the liquidus temperature. To avoid such situations, a low melting point precursor e.g., Al–Si, Al–Cu or Al–Mg, with eutectic composition is prepared with dispersion of the blowing agent without decomposition, and subsequently added to the high temperature melting liquid in the second stage. In the second foaming step, therefore, it becomes easy to control the foam structure. Ferrous alloys have also been attempted by this route using tungsten powder as foam stabilizer; the blowing agent may be $MgCO_3$ or $SrCO_3$ [19, 20] which decompose around 1300 ºC giving out CO_2.

Blowing agent as gas source, spray forming. Spray forming is a technique utilized for the production of alloys and composites with fine and equiaxed grain, low oxide content, and low porosity compared to cast alloys. The technique consists of two distinct but integral processes of melt atomization into micron-sized droplets and their subsequent deposition onto a substrate. Metal matrix composites are produced by this technique by injecting carbides or oxides in the spray of droplets. This way, the reinforcement is incorporated in the growing pre-form. In a similar manner, Kelly et al. [2] and Banhart et al.[45] have introduced blowing agents instead of ceramic particulate. Decomposition of blowing agent during deposition led to large amount of gases in the growing pre-form. They used $BaCO_3$ powder for foaming the $CuSn_6$ alloy and SiO_2 for carbon steel [45], respectively. The porosity of around 60% could be attained. However, the pore structure was highly inhomogeneous due to the fact that the deposition layer is always in turbulence condition and growing cells are destroyed, thus impairing the cell wall stabilization process. However, this technique is effective, as low temperature processing in this route leads to increased viscosity at the deposition layer.

Dissolved gas sources. In general, gases are dissolved in liquid metal depending upon the temperature and externally applied pressure. For example, hydrogen is dissolved in Al, Fe and Cu, nitrogen in Fe, whereas oxygen in Ag and Cu. Such metals with absorbed gases in liquid state eventually undergo a eutectic reaction to a two-phase system (gas + liquid) as the temperature is lowered. A co-ordinated effort to entrap the gas during solidification leads to a porous structure [4, 6, 46–48]. It is generally carried out by directionally solidifying the gas-saturated liquid with a controlled cooling, proper external pressure (usually of the order of gas partial pressure at a particular temperature) and solidification velocity [49, 50]. The process parameters have to be chosen such that the gas bubbles do not float to the surface of the melt but remain near the solidification zone and get entrapped in the solid. Ikeda et al. [21] have studied the lotus-type porous stainless steel by the continuous zone melting technique. They have shown that the porosity increases with increasing partial pressure of hydrogen and the pore diameter decreases with solidification front velocity. In this process, generally, large elongated pores are formed, which are oriented in the solidification direction. However, they appear spherical in a cross-section. A typical structure is shown in Fig. 1. Hyun et al. [16] have extensively worked with this technique and have shown that an increase in the solidification velocity from $0.697\ \mathrm{mm{\cdot}s^{-1}}$to $1.185\ \mathrm{mm{\cdot}s^{-1}}$ leads to a decrease in pore diameter (from 100 μm to 50 μm), as well as porosity (29–26%) and an increase in pore number density (25–110) in the lotus-type porous copper. They have also shown that the tensile strength in the direction parallel to pore orientation is 120 MPa, compared to 80 MPa in the perpendicular direction [4]. The process is capable of producing porosities in the range of 5–75%. However, the pore size distribution is generally non-uniform due to concurrent growth of larger and smaller size pores and their coalescence. Based on this process, steels [21], cobalt, chromium, magnesium, copper [4, 6, 49, 50], Ni_3Al [51] and aluminium [46] have been studied.

In addition to the above methods of foaming from liquid metal, foams can also be produced by casting liquid metal in the interstitial space around spherical and randomly stacked space holding materials [1, 5]. The space holding materials can be inorganic [1, 3] or organic granules [52] or hollow spheres. The final product may contain the space holding material depending upon the requirement. However, to remove the space holder by dissolution, burning or leaching, the content of the space holding fillers should be so high that they are interconnected. Materials containing the space holding fillers are known as "syntactic foams". This technique has been used for making foams of Al, Mg, Zn, Pb, and Sn etc. The main advantages of this process is the precise control over the size of the pore, which is close to the size of the space holders used, and capability to produce the predefined geometry of the component.

Aligned pore structures (similar to dissolved gas foaming) have also been made using directional solidification of monotectic or hypermonotectic alloys. It is well known that this type of alloys is made fibrous by direction solidification [53, 54]. However, if the second phase were removed by using some method, an aligned pore structure would form. This process gives rise to very uniform and evenly distributed cylindrical pores because of the second immiscible liquid phase. Monotectic systems such as Al–Bi, Al–In and Cu–Pb are ductile and thus can be shaped by plastic deformation. The process involves fabrication of aligned fibrous structure followed by removal of fibrous minor phases. Yasuda et al. [7] have used electrochemical etching to remove In from hypermonotectic Al–In alloy system. An aligned cell structure having the diameter in the range of 10–20 μm can be obtained. A typical morphological feature of this kind of foam is shown in Fig. 6.

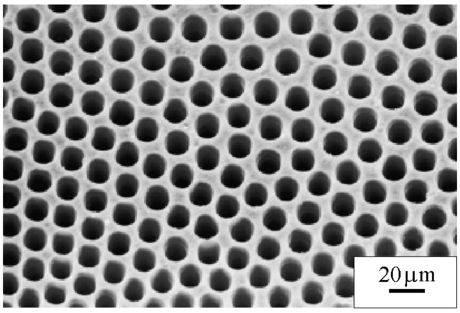

Fig. 6. A uniform distribution of aligned pores (cross -section) obtained from electrochemical etching of directionally solidified monotectic Al–In system [7]

2.2. Powdered metal processing

In the preceding sections, we have discussed the foaming of metallic materials in liquid state. However, solid powdered metal can also be used to make cellular metallic structures. The formation of metallic foam from powdered metal has been extensively studied [1, 14, 17, 32, 55]. Powdered metal can be foamed either by using a blowing agent as gas source or by other techniques where chemical reactions or solid parts are used to create pores during the process. In this section, we will describe these processes in view of the mechanisms of pore formation.

Blowing agent as gas source. Powder compact melting. The powder compact melting process comprises several steps, such as blending of metal or alloy powder with the foaming agent, compaction of powder blend, deformation or working and foaming. Compaction of powders can be accomplished in several ways, such as cold compaction followed by sintering, hot pressing, powder rolling and powder extrusion. The fundamental aim of all such processes up to the foaming step is to form a very dense foamable precursor with uniform distribution of the embedded blowing agent and without any notable residual open porosity. The foamable precursor is subsequently heated to just below the melting point of the matrix material to affect the foaming process [55]. A schematic of the process is shown in Fig. 7.

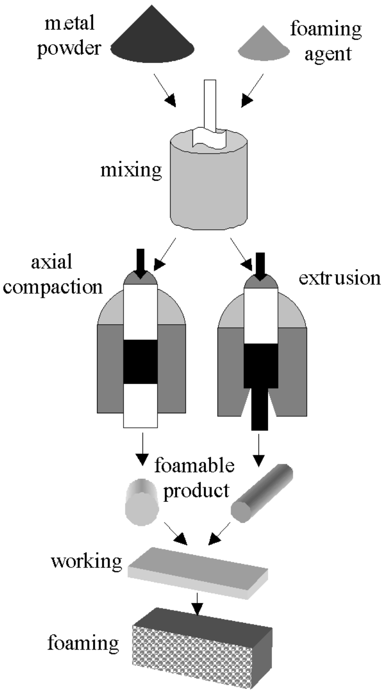

Fig. 7. Processing steps involved in powdered metal foaming with a blowing agent (after ref. [56])

As the temperature is increased, the blowing agent decomposes releasing gas, which in turn leads to expansion of very soft matrix metal. The selection of foaming agent depends upon the sintering and melting temperature of the matrix material. For example, TiH_2 and ZrH_2 are used to foam Al and Zn, respectively [55], $MgCO_3$ and $SrCO_3$ [19] can be used for foaming steels, $BaCO_3$ for copper [2], and $PbCO_3 \cdot Pb(OH)_2$ and MgH_2 for Pb [17]. The decomposition rate, heating rate, and the stability of cell wall structure determine the final density of the foam produced. To delay the decomposition of the foaming agent, it undergoes some heat treatment so that decomposition and melting take place simultaneously. In this way, a uniform foam structure can be developed [57–59]. It has been observed by us that the heating rate of the sample during foaming plays an important role in determining the size and distribution of pores. A relatively high heating rate leads to a smaller pore size with increased pore number density, whereas a slow heating results in larger pore size and the final foam density is higher. This work has been carried out on Al–6.5Si (wt. %) alloy with TiH_2 as the foaming agent. The present authors have also tried out the same composition by the powder rolling technique. The foam structures obtained from these experiments are shown in Fig. 8.

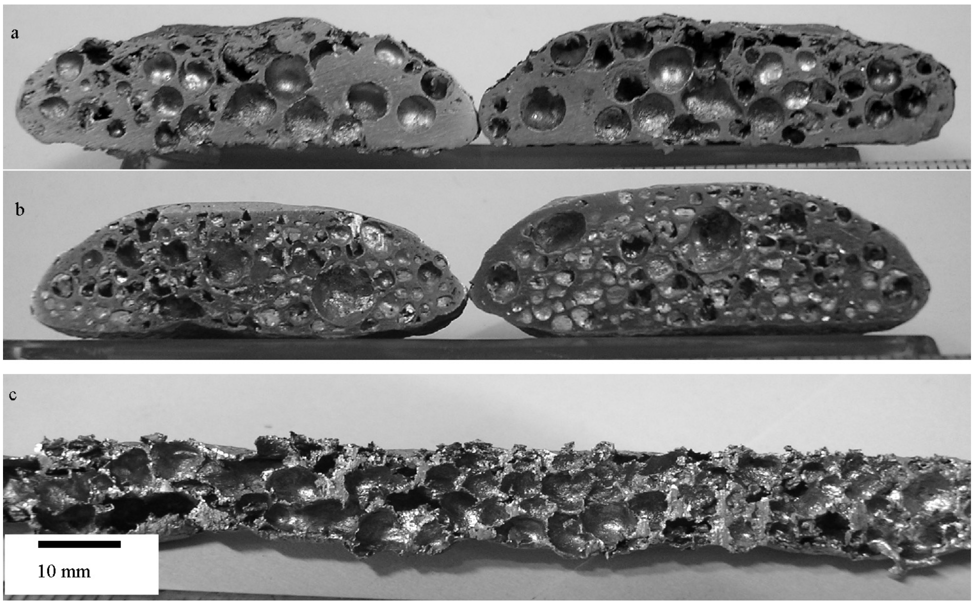

Fig. 8. Aluminium foams by PM route: a) slow heating rate, b) high heating rate and c) by powder rolling technique at high heating rate

Here, it would be worthwhile to mention that the process has been developed where a purposefully incorporated inert gas source is used for the foam genesis instead of a blowing agent [60, 61]. In this process, the material to be foamed is canned and degassed, followed by refilling with an inert gas at a high pressure and sealing. To get a uniform distribution of the entrapped compressed gas in the powder compact, the sealed system is isostatically hot-pressed. Secondary processing of such compacts involves annealing the compact at around $0.6T_m$. The softening of the metal and a simultaneous increasing of pressure of the entrapped gases leads to creeping of the material. As a result, a porous body with 20–40% porosity and 10–100 μm cell size is obtained [60]. The second step of annealing treatment is a long process where foaming is accomplished slowly and, therefore, it takes around 6–24 h for a complete foam formation. Theoretical considerations show that this process cannot give more than 50% porosity [1].

Other foaming routes. As an alternative to the powder compact melting process, mere sintering of metal powders may also be used to produce fine pore structure up to 20–50% porosity. A large variety of materials have been used for this process including titanium, superalloys, bronze and steels. The porosity obtained from this technique consists basically in the interstitial pores present in a powder compact prior to sintering. However, a material that is easily oxidized or produces a reaction product at the surface is not suitable for this process, as the oxide layer inhibits the sintering process. The strength of the foamed component produced via this route is generally very low.

A few processes have been developed where foams are produced by sintering of hollow spheres. There are several reported techniques [30, 62] that can be used to produce hollow spheres. The sintering of hollow spheres can produce both open and closed cell foams. Open structures are obtained by just sintering the hollow spheres, whereas closed cell structures can be generated by filling the interstices between the spheres with a metal powder followed by sintering. This kind of foam is known as syntactic foam. The density of as low as 0.9 $g \cdot cm^{-3}$ has been obtained for Ti–6Al–4V foam produced by this route. The advantage of this process is that the pore size distribution is not random but depends upon the size and distribution of the spheres used, and in this way the properties of the foams become more predictable compared to those having random pore size distribution. In principle, any materials available in powder metallurgy can be used in this route to make foamed parts.

Similar to the above process, IFAM (Bremen, Germany) has recently developed APM (Advanced Pore Morphology) process [63] where spherical foam elements, in the size range of 9–10 mm, have been produced (Fig. 9). The aim of the development was to separate the two steps of foaming and of shaping components. The process involves making a small volume foamable precursor (similar to powdered metal processing) and foaming it in a continuous manner on a conveyor belt furnace. The stack of spherical foamed elements is further sintered to get a component shape. In this way, two types of pore morphology are generated: inner closed porosity of foam elements and the open pores of bulk APM elements. Therefore, it is obvious that this process takes care of hollow sphere sintering as well as foaming by powdered compact melting. The process gives added advantage of open porosity thus decreasing the relative density of aluminium foam to 0.36 $g \cdot cm^{-3}$ compared to other processes. The advantage lies in the fact that the production process has two main steps that are fully independent of each other. They can be carried out at different times, in different locations, and by different companies. The main advantages of APM foam production lies in the fact that all sub-steps in the production of foam elements and components thereof can be fully automated even with the commercially available equipment. If we assume foam element delivery by a commercial supplier, the end-user is not necessarily required to gain any metal foaming know-how or buy cost-intensive high-temperature furnaces.

As we have already discussed the closed cell foam production by the hollow sphere sintering process, if the hollow spheres are replaced by granules of inorganic salts, polymeric materials or metal which can be taken out by various processes such as dissolution, burning or evaporation after sintering, we get an open foam structure. With this technique, Zhao and Sun [9] made aluminium foam using NaCl as the space holding material. The size of the NaCl powder was selected according to the pore size required. Al powder less than 450 μm and NaCl particles around 300–1000 μm were mixed together, compacted and then sintered. The aluminium weight fraction varied between 0.2 and 0.75. Sintering temperature was chosen to be 680 ºC, which is far below the melting point of NaCl (801 ºC) [9]. The sintered samples were kept in warm running water for 45 min to get a porous structure with a relative density of 0.2 and 0.7 for aluminium weight fraction of 0.2 and 0.75, respectively. With the same technique, Zhao et al. [11] tried out the lost carbonate sintering process, where open cell copper was foamed with K_2CO_3 with the melting point of 891 ºC and decomposition temperature of 950 ºC. In this study, both the dissolution and decomposition routes were explored depending upon the production time and shrinkage concerns. The porosity of around 85% could be achieved in this study.

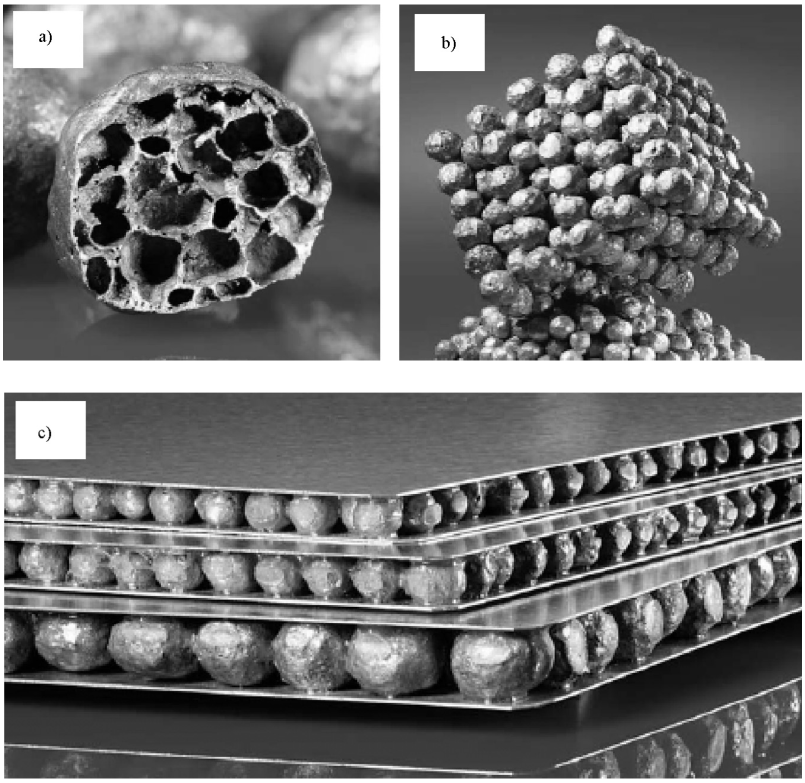

Fig. 9. Illustration of the IFAM (Bremen, Germany) produced: a) foamed spherical elements, b) pure foam parts, c) sandwich structure with a cellular core structure [63] (courtesy: IFAM, Bremen, Germany)

Metallic foams can also be produced by foaming metal powder slurries with an addition of chemicals, which led to the formation of gases as reaction product of the additives. A mixture of metal powder and chemicals are poured into a mould to form viscous slurry and kept there at an elevated temperature. The gases produced by the reaction lead to foaming. A controlled composition of reactants gives rise to stable porous slurry that is dried and sintered subsequently. Aluminium foams have been produced by this route using orthophosphoric acid and hydrochloric acid as additives [64]. Bleck has recently developed steel foam using laminate silicate as dispersant for stabilising the slurry and alcohol solution containing orthophosphoric acid as additives [65]. This process has been named as SlipReactionFoamSintering or SRSS (SchlickerReaktionsSchaumSinter). This process has been utilised for iron and steel powders, 316 Stainless Steel, Hastealloy B and Inconel 625 [65].

3. Stabilisation of foam structure

In the previous sections, we have dwelt mainly on different processing routes for metallic foam. We have shown that the complete foaming process is concerned with foam inception or genesis where the blowing agent decomposes (1), its evolution and growth of spherical pores (2), further foaming that leads to thinning of cell wall and thus to change of pore shape from spherical to polygonal (3), and pore coalescence due to surface tension and gravitational pull (4). This means the eventful life of foams comprises birth, life and death. Our aim is to avoid the death or collapse of foam during the processing. The main convict for the death of foam is the instability of cell walls under the pressure differences or gravity. As we can see from the third and fourth steps, the death or collapse of foam occurs at the peak of third and start of the fourth step [35]. This part of the foaming process has constituted a tremendous challenge for the researchers. The stability of foam can be understood as the formation of a uniform distribution of large size pores, which in turn can be defined as foamability also. If we consider the physics of foaming, the growth of pores or cells lead to the thinning of cell walls and their consequent rupture. This is generally attributed to the drainage of liquid through the cell walls under gravity and also due to surface tension that leads to a positive pressure at the cell wall compared to the triple junctions (Fig. 10a). The pressure at triple junctions of cell walls is lower than that in the cell wall and the liquid on the cell wall tends to flow towards triple points during the foaming process. With the growth of cell size, gravity also comes into effect, and cell walls act as tunnels for the drainage of liquid. This leads to the final breaking of cell wall resulting in coarsening of pores. The drainage phenomenon has been of real scientific importance to understand the physics of the foaming process and efforts have also been made to model aqueous foams [66]. Therefore, it is obvious from the above that strengthening the cell walls can inhibit this phenomenon. This can be accomplished in several ways, e.g., by enhancing the viscosity of melt, using different alloying elements in the melt or employing ex- or in-situ particles that stabilize the wall [27, 40, 41, 67–70]. The breaking of cell walls becomes pronounced, due to pressure differences, when pores with a wide range of size are formed. Cell wall rupture also takes place due to changing rheological characteristics, and deformation and flow of foams. The effect of ceramic particles on the stability of foams has already been discussed in the previous sections. A detailed description is given by Babcsan et al. [41] and Deqing and Ziyuan [40], where they have discussed various aspects of particle reinforcement, such as particle materials, particle size and volume fraction, on foamability of Al alloys by the liquid metallurgy route. Similarly, Koerner et al. [68] discussed the aluminium foam stabilization by the presence of an oxide network. They have also suggested that the effect of the second phase on the stability of foam is due to its wetting characteristics with aluminium liquid. A good wetting (for example, Al/Al_2O_3 contact angle 63^o at 1100 °C) leads to a decrease in the pressure in particle-free regions of the cell wall which in turn reduces the tendency of liquid drainage and thus cell wall thinning. The ceramic particles are segregated at the cell boundaries, and as the cell grows particles are pushed away. Finally the particles mainly decorate the cell walls leading to its strengthening (Fig. 10b).

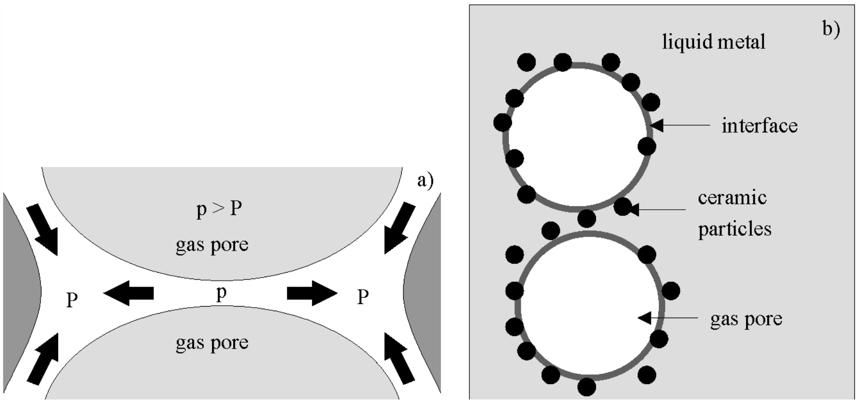

Fig. 10. Cell wall structure and related phenomena during foaming process (a) (after ref. [35]); the effect of ceramic particles for stabilization during the foaming process (b) (after ref. [36])

Sometimes, a pre-treatment of the blowing agent also gives better foaming action [57–59]. A prolonged heating of TiH_2 at different temperatures leads to the formation of an oxide layer on the surface delaying the reaching of decomposition temperature during foaming. Matijasevic and Banhart [58] reported that the delay can be up to 45 s, final expansion of the foam is increased from 4.5 to 5.5 times and a more uniform foam is formed with smoother cell walls and increased roundness of pores. Kennedy and Lopez [71] and Zeppelin et al. [55] have studied the decomposition behaviour of as received and oxidized TiH_2 powder and desorption behaviour of hydrides such as MgH_2, TiH_2 and ZrH_2 and their effect on foaming process.

The drainage phenomenon can also be subdued by foaming in a low gravity environment [33–36]. Foaming in low gravity delays the collapse of cells as the mass of the foam above the cell becomes negligible. The coalescence phenomena become more pronounced particularly when the size of the foaming material is large. Therefore, experiments have been carried out in microgravity conditions where a notable foam expansion is observed also for the material known to be unstable in the presence of gravity but coalescence creates very coarse bubbles. From the different structures of the solidified samples, it was concluded that the presence of oxide particles in the melt hindered cell wall drainage by increasing the viscosity of the melt, and also suppressed the coalescence of the pores. Recently, Babcsan et al. [72] have demonstrated the feasibility of in-situ studies of liquid metal foaming. Otherwise, it would have been difficult to carry out liquid foaming in low gravity parabolic flights or drop towers. The Indian space research program has also highlighted the microgravity experiments and efforts are being made to study such systems in parabolic flight under microgravity conditions.

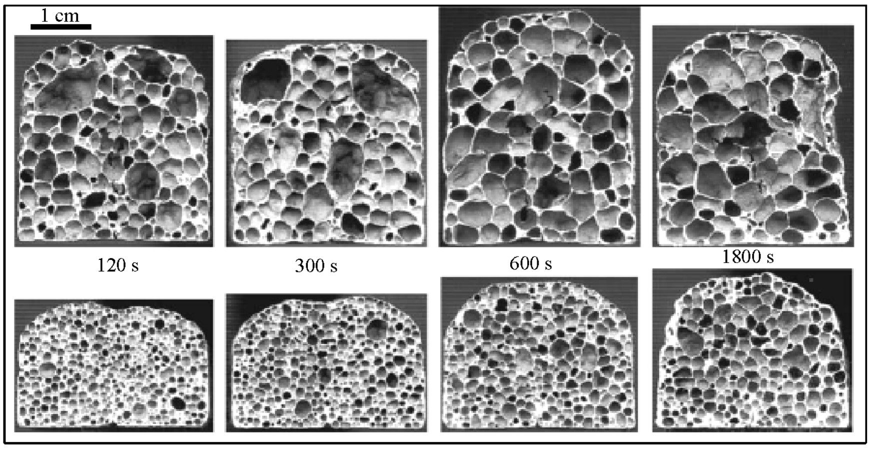

Fig. 11. Foam stability: Cell structure after heating to 665 ^{0}C and increasing the dwelling time at this temperature before solidification (upper row: 1000 mbar argon; lower row: 2800 mbar argon; precursor material: commercial AlSi10Mg0.5, extruded) [68]

In addition to the above foam stabilization techniques, the heating rate during foaming and compaction conditions also influence the final foam characteristics as discussed in detail by Duarte and Banhart [70]. According to them, hot compaction temperature for aluminium with TiH_2 as foaming agent should vary between 400–450 ºC. A higher temperature leads to loss of hydrogen due to early decomposition of the foaming agent, whereas lower temperature of compaction results in improper sintering leaving interconnected pores, and, consequently, in the escape of hydrogen prior to foaming. A high heating rate gives better foaming compared to slow heating. This is attributed to the sufficient time available for the foaming agent to decompose and escape at slow heating rate, and oxidation of a thick oxide layer on the surface as well as inside the precursor material, which inhibits the foaming process by mechanically hindering foam expansion. Koerner et al. [68] experimented with varying ambient argon pressure on aluminium foaming for different time periods. It has been concluded from their study that foaming for 600 s at 665 ºC and at ambient pressure of 1000 mbar does not lead to any drainage. When kept for 1800 s, gas loss was seen only from the top surface of the foam and material redistribution processes induced by the energetically unfavorable cell structures do not take place. A similar experiment at 2800 mbar led to an increase in foam density and decrease in cell size. The observed stability of foams seemed to be independent of the cell size (Fig. 11).

It has also been suggested by Frei et al. [73] that a prior deformation of precursor materials is beneficial to achieve high porosity levels with more uniform cell structures, and also an accelerated foaming. They observed that the as-cast precursor contained unreacted TiH_2 particles within the cell walls, which is not common in deformed precursors. The better stability of the foam has been attributed to the redistribution of the blowing agent within the precursor and deformation-induced damage of the oxide layer on the hydride particles. And also, the pore-free precursor after deformation leads to more dispersed nucleation of pores and prevents their early coarsening.

4. Industrial applications and future prospects

The projections of the future fuel crisis, the requirements of achieving high fuel efficiency along with higher passenger safety in the automobile industry, and the needs of creating light weight construction materials have attracted tremendous consideration for ultralight weight metallic foams. The implementation of metallic foams in these industrial sectors depends to a large extent upon their manufacturing cost, the environmental durability, and fire retardancy [74]. In general, metallic foams possess a range of thermomechanical properties that suggest their application in areas demanding impact/blast amelioration, heat dissipation, acoustic isolation and heat exchange [60, 75]. The projected applications of metallic foams have already been recognised

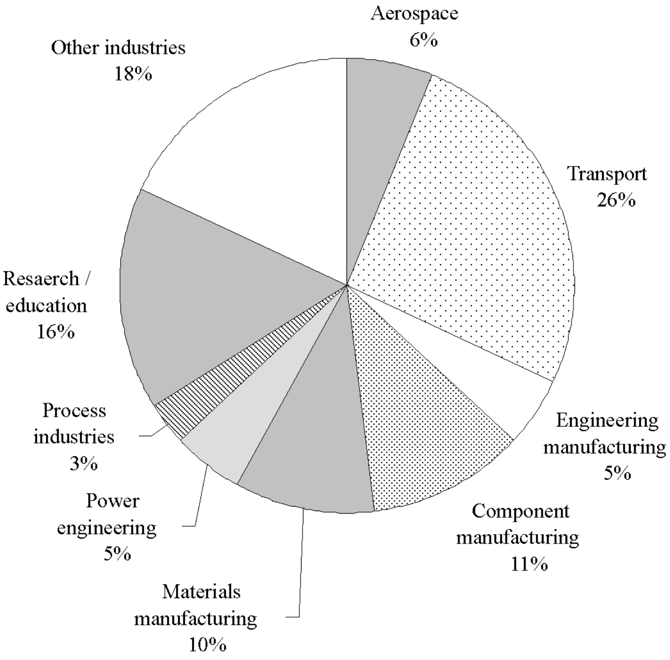

Fig. 12. Break-up of projected industrial sector requirements for metallic foam applications

and a detailed description of the requirements of various industrial sectors can be seen elsewhere [1, 12–15, 76]. Recently, National Physical Laboratory (UK) made a survey both in industries and in research institutions to have an idea of potential expectations from metallic foams. The survey indicated that the major applications of metallic foams, around 32%, in the near future would be directed towards automotive and aerospace industries (Fig. 12).

In addition to this, 26% of the demand would go to materials manufacturing, engineering manufacturing, and component manufacturing. As the transport industry will play a major part in the implementation of metallic foams, we have considered making an attempt to see where we need to use these foams in a car, for example. For this purpose, we should have an idea of the frequent accidental possibility in different parts of a car. We also know that foams in the transport industry are basically needed for weight saving, impact absorbing and thermal insulation. Figure 13 shows a sample car (designed by Karmann, Germany) with components made of Al foam. It is indicated in the figure that 67% of in-vehicle injury cost incurs when the collision takes place at the front end of the car and about 22% in-vehicle injury cost occur during collision from the side. This indicates that the front-end collision of a car is most detrimental during an accident and therefore the utilisation of foamed materials in these parts becomes necessary for passenger safety. As a result, the imperative should be to use metallic foams in the front end for crash safety and the rear end should be replaced with foam to reduce the weight. The difference between these two applications will be based on the quality of the foams to be used in these parts. Al foams have been sought to constitute the largest fraction of components in the automotive industry.

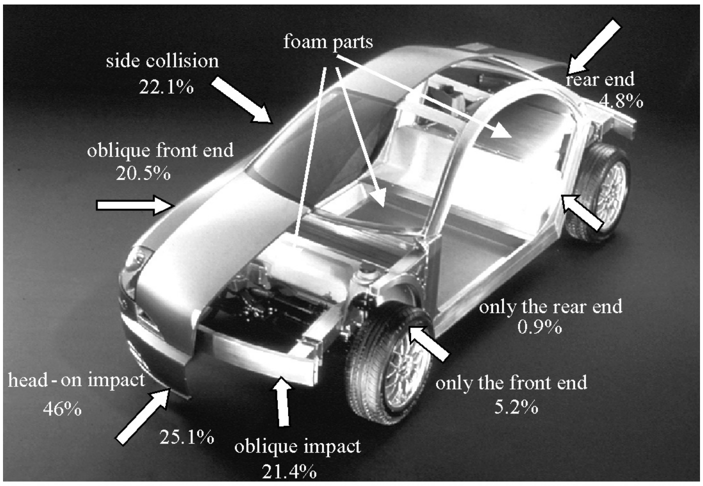

Fig. 13. Karmann car with parts made of aluminium foam (courtesy: IFAM, Bremen, Germany) along with the collision types and subsequent in-vehicle injury cost

The strength of Al foams and the weight saving resulting from their use are currently believed to be insufficient for critical areas of applications. Therefore, tremendous efforts have been recently directed towards steel foam development which, in principle, can have several potential advantages over aluminium foams, such as increased strength and specific stiffness, lower raw material cost, higher melting temperature, and compatibility with steel structures [19]. Initial developments in the processing of steel foams have already been done by conventional powder metallurgical routes, and also by space holding techniques. However, we have a long way to go to produce steel foams commercially, as the high operating temperature and the low viscosity of liquid steels are the major challenges to overcome. The authors of this paper are also actively engaged in the development of steel foams along with the process optimisation for Al foam production.

Similar to the development of steel foams, efforts have been made to develop bulk amorphous metal foams (AMFs). The amorphous foams are seen to have promising application prospects, such as sporting goods and cellular telephone housings due to their high strength, high eleastic strain, and high wear and corrosion resistance. The high corrosion resistance of AMFs compared to other metallic foams is advantageous in applications such as filters, electrodes, and catalyst supports [77–79]. As far as applicability in transport industry is concerned, AMFs may not prove to be a better option than metallic foam, despite its very high strength. Stronger foams may be able to mitigate heavy impacts but simultaneously transmit unacceptably large forces to delicate objects or human beings. Therefore, AMFs may be good for structural applications but not for the transport industry. The processibility and strength of metallic foams are determined by the processing temperature and strength of the parent material. For example, Al alloy-based foams can be easily fabricated by various processes but their mechanical properties are limited by the base alloys, while foams made of stronger crystalline alloys generally require higher-temperature and/or more difficult processing methods. Meanwhile, amorphous metal foams are easy to fabricate due to their low melting temperature and superplastic-like flow at higher temperatures [80]. Despite a low processing temperature, the application of such AMFs above the glass transition temperatures, where toughness is drastically reduced, is limited to 27–480 ºC for an extended period. However, the processing of bulk AMFs is still a challenge due to the compositional accuracy and cooling rate requirements. A slight deviation in composition, due to external additives for foaming, may lead to deviation from the deep eutectic composition resulting in stable crystalline phases. Even if the purity is overcome, the compatibility of external gas blowing phases or space holders and the processing equipment must be ensured to produce amorphous foams. The progress in the identification and understanding of new glass-forming systems with still higher robustness against contaminants and slow cooling, which is currently the subject of intensive research, should lead to an equivalent progress in AMFs [77].

Despite the current development and future possibilities in processing and application of metallic foams, there are several limitations that have to be considered and addressed so as to open new avenues for the metallic foams. The primary hindrance for the wide applicability of metallic foams is the cost of production. The second concern is the variability in materials and the large scatter in measured properties [81]. And also, reliable test methods have to be invented to fully exploit the potential of these cellular materials. World-wide efforts are being made to model the mechanical behaviour of metallic foams, and researchers are, to some extent, successful in predicting the same. However, theoretical and numerical models related to the processing of metal foams, which could be useful in predicting the effect of process parameters, are not available. The developmental process could be made faster by drawing up process models, otherwise "trial and error" would be the only way. The lack of process models is due to the fact that the mechanism of the foaming process is still not fully understood.

5. Concluding remarks

Recent technological advances in the field of metallic foams have led to the development of a wide range of processing techniques for the open, as well as closed cell morphologies. The processing route has to be decided on the basis of the cost of production, materials properties, and the intended applications of the final product. However, as the cost of production of a good quality metallic foam is still high, continual innovations are underway so as to draw maximum benefit by opening new avenues for a wide range of industrial applications. Foams of novel, emerging materials, such as high temperature melting materials, superalloys, steels, bulk metallic glasses etc. are seeing tremendous application opportunities in the near future. However, the lack of full understanding of the mechanisms of foam formation and their theoretical/numerical models inhibits the developmental process to a large extent. A versatile predictive model for processing as well as property has become a necessity for a faster growth in this direction, as well as innovations in processing techniques.

References

[1] Banhart J., Prog. Mater. Sci., 46 (2001), 559.
[2] Kelly P., Wong C.R., Moran A., Int. J. Powder Metall., 29 (1993), 161.
[3] Banhart J., Aluminium, 75 (1999), 1094.
[4] Hyun S.K., Murakami K., Nakajima H., Mater. Sci. Eng. A, 299 (2001), 241.
[5] Rabiei A., O'Neill A.T., Mater. Sci. Eng. A, 404 (2005), 159.
[6] Simone A.E., Gibson L.J., J. Mater. Sci., 32 (1997), 451.
[7] Yasuda H., Ohnaka I., Fujimoto S., Sugiyama A., Hayashi Y., Yamamoto M., Tsuchiyama A., Nakano T., Uesugi K., Kishio K., Mat. Lett., 58 (2004), 911.
[8] Stöbener K., Baumeister J., Rausch G., Busse M., Metal Powder Rep., Jan. (2005), 12.
[9] Zhao Y.Y., Sun D.X., Scripta Mater., 44 (2001), 105.
[10] Chou K.S., Song M.A., Scipta Mater., 46 (2002), 379.
[11] Zhao Y.Y., Fung T., Zhang L.P., Zhang F.L., Scipta Mater., 52 (2005), 295.

[12] DAVIS G.J., ZHEN S., J. Mater. Sci., 18 (1983), 1899.
[13] BANHART J., Ind. Foundry Jo., 51 (2005), 36.
[14] BANHART J., MRS Bulletin, April (2003), 290.
[15] BANHART J., WEAIRE D., Phys. Today, July (2002), 37.
[16] HYUN S.K., NAKAJIMA H., Mat. Lett., 57 (2003), 3149.
[17] IRRETIER A., BANHART J., Acta Mater., 53 (2005), 4903.
[18] PARK C., NUTT S.R., Mater. Sci. Eng. A, 299 (2001), 68.
[19] PARK C., NUTT S.R., Mater. Sci. Eng. A, 288 (2000), 111.
[20] PARK C., NUTT S.R., Mater. Sci. Eng. A, 297 (2001), 62.
[21] IKEDA T., AOKI T., NAKAJIMA H., Met. Mat. Trans., 36A (2005), 77.
[22] HODGE A.M., DUNAND D.C., Intermetall., 9 (2001), 581.
[23] BANHART J., BELLMANN D., CLEMENS H., Acta Mater., 49 (2001), 3409.
[24] WEN C.E., MABUCHI M., YAMADA Y., SHIMOJIMA K., CHINO Y., ASAHINA T., Scripta Mater., 45 (2001), 1147.
[25] MAINE E., ASHBY M.F., Adv. Eng. Mater., 2 (2000), 205.
[26] MIYOSHI T., ITOH M., AKIYAMA S., KITAHARA A., Proc. MRS Symp. D. S. Schwartz, D. S. Shih, A. G. Evans, H. N. G. Wadley (Eds.), 521 (1998), 133.
[27] IP S.W., WANG Y., TOGURI J.M., Can. Met. Quart., 38 (1999), 81.
[28] DEQING W., ZIYUAN S., Mater. Sci. Eng. A, 361 (2003), 45.
[29] LLYOD D.J., MCLEOD A.D., MORRIS P.L., JIN I., Int. Patent (1991), WO 91/19823.
[30] WAAG U., SCHNEIDER L., LOETHMANN P., STEPHANI G., Metal Powder Rep., 55 (2000), 29.
[31] WADA T., INOUE A., Mater. Trans., 44 (2003), 2228.
[32] JIANG B., ZHAO N.Q., SHI C.S., DU X.W., LI J.J., MAN H.C., Mater. Lett., 59 (2005), 3333.
[33] WEAIRE D., Adv. Eng. Mater., 4 (2002), 723.
[34] BANHART J., BAUMGAERTENER F., COX S.J., KRONBERG B., LANGEVIN D., ODENBACH S., WEAIRE D., WUEBBEN T., Proc. 1st Int. Symp. Microgravity Research and Applications in Physical Sciences and Biotechnology, ESA Publications, ESTEC, Noordwijk, 2001, p. 589.
[35] WUEBBEN T., ODENBACH S., BANHART J., Proc. Eurofoam 2000 P. Zitha, J. Banhart, G. Verbist, (Eds.), MIT-Verlag Bremen, 2000, pp. 98–103.
[36] WUEBBEN T., STANZICK H., BANHART J., ODENBACH S., J. Phys.: Cond. Matter, 15 (2003), S427.
[37] JIN I., KENNY SANG L.D., NA H., US Patent 4,973,358.
[38] ASHOLT P., Proc. Int. Conf. Bremen, Germany, June 14–16, 1999, J. Banhart, M.F. Ashby, N. Fleck, (Eds.), MIT Press-Verlag, p. 133.
[39] BANHART J., JOM, 52 (2000), 22.
[40] DEQING W., ZIYUAN S., Mater. Sci. Eng. A, 361 (2003), 45.
[41] BABCSAN N., LEITMEIER D., DEGISCHER H.P., Mat.-Wiss. U. Werkstofftech, 34 (2003), 22.
[42] PRAKASH O., EMBURY J.D., SANG H., SINCLAIR C., SILVETTI P., *The Minerals, Metals and Materials Society,* C.M. Ward-Close, F.H. Froes, D.J. Chellman, S.S. Cho (Eds.), 1997. p. 19.
[43] SIMONE A.E., GIBSON L.J., Acta Mater., 46 (1998), 3109.
[44] MA L., SONG Z., Scripta Mater., 39 (1998), 1523.
[45] BANHART J., KNUEWER M., Proc. PM 1998 World Congress, European Powder Metallurgy Association, 5 (1998), 265.
[46] BONENBERGER R.J., KEE A.J., EVERETT R.K., MATIC P., Proc. MRS Symp., D.S. Schwartz, D.S. Shih, A.G. Evans, H.N.G. Wadley (Eds.), 521 (1998), 303.
[47] APPRILL G.M., POIRIER D.R., MAGUIRE M.C., GUTSCH T.C., Proc. MRS Symp., D.S. Schwartz, D.S. Shih, A.G. Evans, H.N.G. Wadley, 521 (1998), 291.
[48] TANE M., ICHITSUBO T., NAKAJIMA H., HYUN S.K., HIRAO M., Acta Mater., 52 (2004), 5195.
[49] NAKAJIMA H., HYUN S.K., OHASHI K., OTA K., MURAKAMI K., Colloids Surf., A179 (2001), 209.
[50] YAMAMURA S., SHIOTA Y., MURAKAMI K., NAKAJIMA H., Mater. Sci. Eng. A, 318 (2001), 137.
[51] HYUN S.K., NAKAJIMA H., Mater. Sci. Tech., 74 (2004), 667.

[52] MA L., SONG Z., HE D.P., Scripta Mater., 41 (1998), 785.
[53] GRUGEL R.N., HELLAWELL A., Metall. Trans. A, 12 (1981), 669.
[54] DHINDAW B.K., STEFANESCU D.M., SINGH A.K., Metall. Trans. A, 19 (1988), 2839.
[55] ZEPPELIN F., HIRSCHER M., STANZICK H., BANHART J., Composite Sci. Tech., 63 (2003), 2293.
[56] BANHART J., BAUMEISTER J., J. Mater. Sci., 33 (1998), 1431.
[57] MATIJASEVIC B., FIECHTER S., ZIZAK I., GÖRKE O., WANDERKA N., SCHUBERT-BISCHOFF P., BANHART J., Powder Metallurgy World Congress, H. Danninger, R. Ratzi (Eds.), European Powder Metallurgy Association, Vienna, 4 (2004), 149.
[58] MATIJASEVIC B., BANHART J., Scripta Mater., 54 (2006), 503.
[59] MATIJASEVIC B., BANHART J., FIECHTER S., GÖRKE O., WANDERKA N., Acta Mater. 54 (2006), 1887.
[60] KEARNS M.W., BLENKINSHOP P.A., BARBER A.C., Int. J. Powder Metall., 24 (1998), 59.
[61] MURRAY N.G.D., SCHUH C.A., DUNAND D.C., Scripta Mater. 49 (2003), 879.
[62] SYPECK D.J., PARRISCH P.A., WADLEY H.N.G., Proc. MRS Symp. D.S. Schwartz, D.S. Shih, A.G. Evans, H.N.G. Wadley (Eds.), 521 (1998), 205.
[63] STÖBENER K., BAUMEISTER J., RAUSCH G., Metal Powder Rep., Jan. (2005), 12.
[64] DROLET J.P., Int. J. Powder Metall., 13 (1977), 223.
[65] BLECK W., Metal Powder Rep., March (2004), 18.
[66] GERGELY V., CLYNE T.W., Acta Mater., 52 (2004), 3047.
[67] WASAN D.T., NIKOLOV A.D., LOBO L.A., KOCZO K., EDWARDS D.A., Progr. Surface Sci., 39 (1992), 119.
[68] KOERNER C., ARNOLD M., SINGER R.F., Mater. Sci. Eng. A, 396 (2005), 28.
[69] ESMAEELZADEH S., SIMCHI A., LEHMHUS D., Mater. Sci. Eng. A, 424 (2006), 290.
[70] DUARTE I., BANHART J., Acta Mater., 48 (2000), 2349.
[71] KENNEDY A.R., LOPEZ V.H., Mater. Sci. Eng, A, 357 (2003), 258.
[72] BABCSAN N., GARCIA-MORENO F., LEITLMEIER D., BANHART J., Mater. Sci. Forum, 508 (2006), 275.
[73] FREI J., GERGELY V., MORTENSEN A., CLYNE B., Adv. Eng. Mater., 4 (2002), 749.
[74] BEALS J., THOMPSON M., J. Mater. Sci., 32 (1997), 3595.
[75] LU T.J., STONE H.A., ASHBY M.F., Acta Mater., 46 (1998), 3619.
[76] BANHART J., Proc. 9th Int. Conf. on Aluminium Alloys J.F. Nie, A.J. Morton, B.C. Muddle (Eds.), 2004, p. 764.
[77] JOHNSON W.L., Mater. Sci. Forum, 225 (1996), 35.
[78] INOUE A., Acta Mater., 48 (2000), 279.
[79] BROTHERS A.H., DUNAND D.C., Scripta Mater., 54 (2006), 513.
[80] WANG W.H., DONG C., SHEK C.H., Mater. Sci. Eng. R, 44 (2004), 45.
[81] RAMAMURTY U., PAUL A., Acta Mater., 52 (2004), 869.

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