Full length article

Microporosity in aluminium foams

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ARTICLE INFO

ABSTRACT

We studied microporosity in the metallic matrix of aluminium foams produced by the powder metallurgical route both with and without application of a blowing agent. Microporosity was studied in-situ in liquid metal foams as well as ex-situ in the solidified microstructures. In-situ studies were carried out using synchrotron X-rays. Quantitative analyses of the amount and distribution of microporosity inside cell walls, Plateau borders and nodes were performed on 2D micrographs and on 3D reconstructed volumes generated by X-ray tomography. We studied the influence of alloying elements, blowing agent and holding time on the amount and type of micropores. The mechanisms of microporosity formation and the evolution of microporosity via diffusion of hydrogen and by coalescence are discussed. It was observed that alloy composition and holding time have a strong influence on microporosity. Different possible strategies to control microporosity are suggested.

1. Introduction

Microporosity is inevitable in aluminium alloys [1–4]. The amount of microporosity is determined by the solidification conditions and alloying elements. While insufficient feeding during solidification leads to shrinkage microporosity, the difference in hydrogen solubility between liquid and solid aluminium results in gas microporosity [2]. Microporosity adversely affects properties such as tensile [1,4] and fatigue strength [3] because it promotes stress concentration. Foams made from aluminium alloys also contain such microporosity. We study it in this article.

At present, the disparity between the predicted and measured strength of aluminium foams is attributed to the presence of a non-uniform cell size distribution and defects such as missing or broken cell walls, elliptical cells and the curvature of cell walls [5–7]. However, the effect of microporosity is not taken into account. It has been reported that castings of thin sections are more vulnerable to the effects of micropores because they reduce the load-bearing cross section considerably [1]. The same also applies to closed-cell metal foams, which are made of thin sections such as

cell walls and Plateau borders, usually with thicknesses of 50−300 μm. This implies that the tensile and fatigue properties of closed-cell metal foams are also affected by microporosity.

According to the most simple model, the plastic collapse stress of foams scales with the relative density of foams as follows [8]:

$$\frac{\sigma_{pl}^*}{\sigma_{ys}} = 0.3 \times \left(\phi \frac{\rho^*}{\rho_s} \right)^{3/2} + 0.4 \times (1-\phi)\frac{\rho^*}{\rho_s}, \tag{1}$$

where ρ^* and σ_{pl}^* are the density and plastic collapse stresses of the foam, ρ_s and σ_{ys} are those of the solid cell wall material. ϕ is the volume fraction of solid contained in the cell edges and the remaining fraction $(1-\phi)$ is in the cell faces. In Eq. (1), the first term is for the bending of cell edges (Plateau borders) and the second term is for the stretching of cell faces (cell walls). A similar relationship that includes bending and stretching components also exists for Young's modulus. While stretching immediately implies tensile stresses, bending involves both tensile and compressive stresses. Indeed, it is shown from in-situ compression tests of aluminium foams that even under compressive loading, tensile stress is generated [9]. Therefore, beside under tensile and fatigue loading, metal foams are expected to be susceptible to the effects of microporosity under compressive loading as well. While ρ^* in Eq. (1) takes into account the total porosity of the foam it does not

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provide any clue about the amount of microporosity. For example, a foam with $\rho* = 0.18$ has 82% porosity and 18% solid fraction. Let us assume that out of this 82% porosity 2% is present in the form of microporosity in the microstructure of the solid part and the remaining 80% porosity is related to the volume of the cells. This implies that the microstructure of the cell walls, Plateau borders and nodes contains 10% microporosity. While the influence of 2% porosity on the mechanical properties of a foam is marginal, the effect of 10% microporosity on the tensile strength of the metallic constituent cannot be ignored [10]. This implies that the relative density alone cannot predict the mechanical properties of a foam and the effect of microporosity should also be considered.

Ohgaki et al. [11] and Toda et al. [9,12] were the first to draw attention towards the effect of microporosity in aluminium foams. A high level of microporosity (26% of the metal volume) was reported [11]. It was shown that in metal foams subjected to compressive stress, cracks originate from micropores with diameters between 30 μm and 350 μm. This is because large strains accumulate at such micropores and therefore the borders of them can be crack initiation sites [11]. Our own previous study has also shown that under compression of foams cracks are generated at micropores of deformed cell walls [13].

The goal of the present work is to quantify the amount and type of microporosity in aluminium alloy foams and clarify the mechanisms of microporosity formation. We also propose strategies to modify foams in Section 4.6.

2. Experimental

2.1. Materials

Aluminium (Alpoco, 99.7% pure, $D_{50} = 38$ μm), silicon (Wacker Chemie, 99.5% pure, $D_{50} = 26$ μm), copper (Chempur, 99.5% pure, $D_{50} = 27$ μm), pre-alloyed AlMg50 (Possehl Erzkontor GmbH, purity not specified) and TiH_2 serving as blowing agent (Chemetall, Grade N, 98.8% pure, $D_{50} = 14$ μm) powders were used to prepare foamable precursors following the powder metallurgical (PM) route. D_{50} is the value of the particle diameter at 50% in the cumulative distribution. The TiH_2 powder was heat-treated at 480 °C for 180 min in air. To prepare the precursors, 30 g of metal powder were mixed with or without addition of 0.5 wt% of TiH_2 powder in a tumbling mixer for 15 min. The powder blend was subjected to uni-axial compaction in a cylindrical die at 400 °C for 5 min applying a pressure of 300 MPa. Four alloys were prepared with TiH_2 and two alloys without TiH_2 as specified in Table 1. Alloy AlSi6Cu4 (values in wt.%) was prepared both with and without TiH_2. Unless otherwise indicated, the version containing TiH_2 is meant. AlCu13Mg4 was prepared only without TiH_2. Cylindrical tablets (36 mm diameter, ~11 mm thickness) were obtained by uni-axial compaction. $10 \times 10 \times 4$ mm^3 large samples were cut out from these tablets for foaming, ensuring that the compaction direction was along the 4-mm long side of the sample.

2.2. Foaming procedures

Two different furnaces were used for foaming. All samples containing TiH_2 were foamed inside a steel mould in air using a *lamp furnace* equipped with two or three halogen lamps of 150 W power [14], see also Ref. [15]. The temperature was measured at the bottom surface of the sample. TiH_2-free samples were foamed by gas pressure manipulation in a gas-tight *pressure furnace* equipped with a ceramic heater. The temperature was measured at the bottom surface of the sample and calibrated to extract the temperature of the foam interior (given in Table 1), see Ref. [15].

TiH_2-containing samples were foamed by heating them to above their melting point at a heating rate of 2–3 K/s. After the temperature had reached the foaming temperature it remained at or slightly above that level for a period which is denoted as *holding time* (HT). After holding, the foam was solidified by ambient cooling at an average cooling rate of initially about 1 K/s. To foam TiH_2-free samples, the powder compacts were first heated up to the foaming temperature inside the pressure furnace filled with argon at 5 bar pressure. After melting the samples, the gas pressure was released to ambient pressure (1 bar) within 30 s. The pressure drop immediately induced expansion of the sample by the release and expansion of adsorbed gases present in the powder compact. This process is referred to as *pressure induced foaming* (PIF) [16,17]. After pressure release, the experimental course was identical to that of the samples containing blowing agent.

2.3. In-situ observation of foaming

Foaming of Al and AlSi6Cu4 was observed in-situ by radioscopy using hard X-ray synchrotron radiation at beamline ID19 of the European Synchrotron Radiation Facility. A mixed mode of absorption and phase contrast was used for image acquisition that ensures high visibility of the cell walls. In the phase contrast mode, the contrast of the liquid-gas interfaces is dominated by refraction effects. The details of the experimental technique are given in Refs. [18,19]. Briefly, the white radiation from the beamline's wiggler insertion device (gap 40 mm) was used for illumination (mean energy ~34 keV). As detector a 300-μm thick YAG:Ce scintillator was coupled by a $1 \times$ magnifying lens to a high-speed CMOS camera (pco.1200hs, PCO AG, Germany) located several meters downstream of the sample.

2.4. Structural characterization

The solidified foams were sectioned into two halves. One half was used for optical microscopy. The samples were embedded in cold-curing resin (Kulzer), mechanically ground using 120–4000 grit silicon carbide paper, polished successively with 3 μm and 1 μm diamond paste, and finally polished with a SiO_2 suspension on a smooth cloth. *ImageTool* version 3.00 software was used to analyse the microporosity from the micrographs.

From the other half of the AlSi6Cu4, AlSi6Cu4 without TiH_2 and

Table 1
Alloy compositions and foaming parameters. The initial cooling rate in all cases was 1 K/s.

Composition (wt%)	TiH_2 used?	Holding times (s)	Foaming (interior) temperature (°C)	Pressure profile
Al	yes	50	670	constant at 1 bar
AlSi7	yes	200	620	constant at 1 bar
AlSi11	yes	200	620	constant at 1 bar
AlSi6Cu4	yes	200, 600, 1000, 1500, 2500	620	constant at 1 bar
AlSi6Cu4	no	200, 1500	620	5 bar during melting, then reduced to 1 bar
AlCu13Mg4	no	200	620	5 bar during melting, then reduced to 1 bar