8th International Conference on Porous Metals and Metallic Foams, Metfoam 2013

Microstructural evolution in investment casted open-pore aluminum-based alloy foams

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Abstract

In the present study open-pore metal foams of the high purity alloys AlSi7 and AlZn11Mg1 as well as the technical purity alloys AA5019 and AA7050 are manufactured by investment casting. Subsequently, their microstructural evolution is characterized at different distances to the deadhead to illustrate the influence of the depth-dependent cooling conditions.

1. Introduction

Open-pore metal foams possess attractive properties due to their highly porous structure in combination with its base material. Hence, this group of material is of great interest for different fields of application. Although much has been reported about metal foams, the focus was primarily on their manufacturing methods, effective properties (mechanical, thermal, etc.) and characterization of the foam-like structure, but their microstructure has been covered rarely. However the microstructure does have a grave effect on the effective properties for which reason the microstructural evolution in investment casted open-pore Al foams are investigated as follows.

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2. Aim of investigations

Investment casted open-pore metal foams mostly show a microstructure different to other conventional casting methods. On the one hand this is due to the small dimensions of the single strut cross section and on the other hand it is due to the slow cooling velocity of the plaster mold. Especially materials with a low melting point such as aluminum can undergo a partial heat treatment. Hence, the microstructure does in most cases not exhibit a common cast structure. This phenomenon is however depending on the alloy. For this reason, in the present study, different open-pore Al foams are manufactured and subsequently characterized. The characterization of the microstructure is carried out by microscopical methods and by hardness tests. All these methods are applied at three different distances to the deadhead to illustrate the influence of the depth-dependent cooling conditions by comparing it with literature data of near-equilibrium solidification and principal precipitation processes for each alloy. Based on these experiments, the post-casting properties of different alloys can be evaluated and a first prediction of how different open-pore Al foams have to be heat treated to achieve the postulated properties can be deduced.

3. Experimental

3.1. Materials

The starting materials for this work are aluminum alloys in high purity and technical purity. The high purity alloys are AlSi7 (obtained from Aalen University, GER) and AlZn11Mg1 (obtained from Ruhr-University Bochum, GER) and the technical purity alloys are commercial AlMg5 (5019) and AlZn6MgCuZr (7050). Their compositions analyzed by X-ray fluorescence spectrometry is shown in Table 1.

3.2. Processing

Open-pore Al foam samples are in-house fabricated by investment casting as described similarly by Wang et al. (2001) and Yamada et al. (1999). A commercial reticulated open-pore polymer foam (obtained from Reigies Schaumstoffe GmbH, GER) with a pore density of $\rho_p = 10$ ppi is used as preform. By a thermal-additive process (Matz et al. (2014)), this preform is treated to achieve a relative density of $\rho_{rel} \approx 12.5\% \pm 0.2\%$. The preform in its dimensions of $50 \cdot 45 \cdot 20 \text{ mm}^3$ is infiltrated by plaster. In a next step the plaster mold is heated in an incineration furnace at $\vartheta = 360 \, ^\circ\text{C}$ for dewaxing and at $\vartheta = 720 \, ^\circ\text{C}$ to pyrolyze the polymer preform and to strengthen the mold.

The metallurgical processing is carried out by centrifugal casting (cf. Müller et al. (2013)). The starting material is placed in a pre-heated crucible and vacuum melted. At $\vartheta = 880 \, ^\circ\text{C}$, the casting is induced and the mold, which is preheated at $\vartheta \approx 400 \, ^\circ\text{C}$, is infiltrated by the melt. After $t = 10 \, \text{min}$ the mold is removed from the casting machine whereupon the cooling takes place at atmospheric conditions. In a last step the samples are cleaned by water jet and Tetranatriumethylendiamintetraacetate $(\text{C}_{10}\text{H}_{12}\text{N}_2\text{Na}_4\text{O}_8)$.

3.3. Microstructural characterization

Microstructural characterization is investigated at three different sections of the open-pore Al foam samples as shown in Figure 1. Thereby, the impact of the casting process and the alloy composition on the microstructure of the metal foam as function of the cooling process can be determined.

Table 1. Chemical composition of aluminum alloys (mass fraction).

| Alloy         | Si (%) | Fe (%) | Cu (%) | Mn (%) | Mg (%) | Zn (%) | Cr (%) | Ti (%) | Zr (%) |
|---------------|--------|--------|--------|--------|--------|--------|--------|--------|--------|
| AlSi7         | 7.54%  | 0.09%  | 0.05%  | –      | –      | 0.03%  | –      | 0.01%  | –      |
| AlZn11Mg1     | 0.05%  | 0.02%  | 0.02%  | –      | 0.99%  | 11.28% | –      | –      | –      |
| AlMg5         | 0.28%  | 0.39%  | 0.21%  | 0.17%  | 5.32%  | 0.13%  | 0.12%  | 0.12%  | –      |
| AlZn6MgCuZr   | 0.17%  | 0.14%  | 2.42%  | 0.08%  | 2.13%  | 6.25%  | 0.05%  | 0.05%  | 0.13%  |
|               |        |        |        |        |        |        |        |        | Bal.   |
|               |        |        |        |        |        |        |        |        |        |
Metallographic preparation

The samples are partitioned in a distance of \( s_1 = 45 \text{ mm} \) (max. distance to deadhead), \( s_2 = 25 \text{ mm} \) (mid of sample) and \( s_3 = 5 \text{ mm} \) (close to deadhead) by a wet disc grinder. The single sample pieces are embedded in a cold polymerizing plastic (ClaroCit from Struers GmbH, GER) for further mechanical sample preparation. This is carried out by an automatic grinding/polishing machine (TegraPol-21 from Struers GmbH, GER) in two main steps, namely wet grinding and diamond polishing as well as chemico-mechanical polishing. Subsequent to these preliminary steps, electrolytic etching after Barker is applied as described in Petzow (2006).

Metallographic analysis

Metallographic characterization of the prepared samples is obtained by a light-optical microscope (DMI 500 M from Leica GmbH, GER) at amplifications of \( V = 50:1 \) or \( V = 100:1 \) in polarized light for evaluating the microstructural morphology.

3.4. Mechanical Characterization

To achieve further information about the microstructural evolution in the single sections of the metal foam samples, hardness measurements are applied by a Vickers hardness tester (type 38163 from KB Prüftechnik GmbH, GER) with HV1. At each section several measurements are taken and the arithmetic mean is composed.

4. Results and discussion

In the following, the results of the experimental investigations are pointed out, starting with the aluminum alloys in high purity AlSi7 and AlZn11Mg1 and followed by the aluminum alloys in technical purity AlMg5 (5019) and AlZn6MgCuZr (7050).

4.1. Open-pore AlSi7 foam

The hypoeutectic AlSi7 foam shows significant differences in its microstructure which can be characterized by the dendritic formation of the \( \alpha(\text{Al}) \) and the formation of AlSi-eutectic. At the beginning of the solidification, primary \( \alpha(\text{Al}) \) dendrites are formed which grow while the temperature decreases. During this, the Si-concentration increases to up to 1.67% (Mondolfo (1971)). Due to the small solubility of Si in \( \alpha(\text{Al}) \), the residual melt enriches with Si until eutectic composition is accomplished and solidifies with proceeding reduction in temperature. In this alloy, the eutectic is mainly located in the inter-dendritic spacing of the primary and dendritic precipitated \( \alpha(\text{Al}) \) solid solution. This microstructure notably arises in section 3 (close to deadhead). As shown in Figure 2, the \( \alpha(\text{Al}) \) is existent as comparably fine \( \alpha \)-dendrites whereas AlSi-eutectic is mainly located in the spacing of the secondary dendritic arms. In section 2, comparatively coarse dendrites are dominating and there are hardly any typically dendritic arms in contrast to section 3. Section 1 exhibits mainly a microstructure without dendrites in the single
grains. Rather, α(Al) with comprehended second phase in terms of AlSi-eutectic can be identified. The lamellar eutectic is thereby split into several segments which indicates a microstructural change due to a partial heat treatment (Páramo et al. (2000)).

4.2. Open-pore AlZn11Mg1 foam

When solidifying AlZn11Mg1, crystallization of α(Al) is dominating and subsequently formation of the phases τ(Al2Mg3Zn3) and η(MgZn2) out of the residual melt occurs (Valdez et al. (2012)). This can be seen in the optical
micrographs whereby the in a final step solidified phases are visible around dendritic \( \alpha(\text{Al}) \) cells. This kind of microstructural morphology can be detected in section 3 of the open-pore AlZn11Mg1 foam (cf. Fig. 3). In section 2 this kind of network is not present in such a pronounced manner and in section 1 it is hardly evident. This is due to the decelerated cooling at these locations in the sample which leads to balancing of concentration in the microstructure attended with a formation of a fine interconnected network of \( \tau \) and \( \eta \) at the grain boundaries as well as partly within the grains and the grain boundaries get straightened (Brenner et al. (1954)). The thereby occurring decomposition process causes an increase in strength (Wunderlin et al. (1975)) whereby the rise in hardness with increasing distance to the deadhead, shown in Fig. 6, can be explained.

4.3. Open-pore AlMg5 (5019) foam

The single sections of AlMg5 foam mainly differ in terms of the precipitation morphology of the second phase. When solidifying AlMg5 melt, \( \alpha(\text{Al}) \) solid solution containing approx. 1% Mg crystallizes and the residual melt enriches with Mg. Subsequently \( \text{Mg}_2\text{Al}_3 \) precipitates at the grain boundaries (Wen et al. (2005)) or within grains (Hatch (1984)) (due to the presence of other elements in amounts of 0.1-0.4% (cf. Table 1), also \( (\text{Fe,Cr})_3\text{SiAl}_{12} \) and \( \text{Mg}_2\text{Si} \) forms). This microstructure evolves generally in all sections of the sample as shown in Figure 4. In Section 3, \( \alpha(\text{Al}) \) matrix with small precipitations can be seen. In comparison, the microstructure in section 2 shows a higher amount of single precipitations which in section 1 stay stable in quantity but become coarse, this is due to initial aging which gets accelerated by a lower rate of cooling (Dahl et al. (1955)). The hardness profile remains unaffected (cf. Figure 6).

4.4. Open-pore AlZn6MgCuZr (7050) foam

AlZnMgCu alloys stand out due to the highest strength of Al alloys. In contrast to the afore investigated alloys, 7050 exhibits a rather complex character of phases in the \( \alpha(\text{Al}) \) matrix which includes coarse melt crystallizing particles of several micrometers in size, submicron high temperature precipitating phase particles and typical particles due to aging of smaller than 0.1 \( \mu \text{m} \) (Hahn et al. (1975)). The major phases aside \( \alpha(\text{Al}) \) present in the as-cast microstructure are based on \( \eta(\text{MgZn}_2) \), \( \tau(\text{Al}_2\text{Mg}_3\text{Zn}_3) \), \( \text{S(Al}_2\text{CuMg)} \) and \( \theta(\text{Al}_2\text{Cu}) \) (Mondal et al. (2005)). The here studied metal foam samples show a dendritic solidified \( \alpha(\text{Al}) \) matrix (Fig. 5). Amongst the dendritic arms the above named phases are present. It is notable that there are first signs of changes of the configuration and content of the constituents which can be ascribed to partial homogenization treatment (Chen et al. (2003)). Furthermore, an increase in grain size with progressive distance to the deadhead is observable which indicates a minor cooling velocity (Zhang et al. (2012)) in section 1 compared to section 2 and primarily section 3 which is directly associated with the corresponding hardness shown in Fig. 6.

![Fig. 6. Hardness profile as a function of distance to deadhead](image-url)
5. Conclusion

• The microstructural evolution of the as-cast samples strongly depends on the alloy composition and hence on the nucleation and diffusion mechanisms of each element.
• At different sections of the open-pore metal foams, a different microstructural evolution results, which is shown by individual attributes of each alloy and the different hardness profiles as a function of distance to deadhead. Viz. any used alloy for investment casted open-pore metal foams requires an adequate understanding of the microstructural mechanisms.
• The microstructural evolution in all investigated samples shows an indication of a maximum cooling velocity close to the deadhead. Furthermore, it can be concluded that at the beginning of the casting process an increase of mold temperature occurs as also indicated by O’Mahoney et al. (2000). After solidification, cooling of the region close to the deadhead is more pronounced as consequence of the high thermal conductivity $\lambda$ of the Al alloy which results in a better heat dissipation. In contrast, the thermal conductivity of the plaster is much smaller but its heat capacity $c_p$ is much higher which leads to a continuous heat input into deadhead distant regions.
• A more homogenous microstructural evolution throughout the whole open-pore metal foam requires quenching or controlled furnace cooling of the mould directly after casting.

Acknowledgements

The authors gratefully acknowledge the European Union/European Fund for Regional Development and the federal state Baden-Württemberg for financial support. The authors would also like to express their gratitude to M. Bossler for proof reading.

References

Brenner, P., Schippers, M., 1954. Gefügeerscheinungen bei der Entmischung homogenisierter Aluminium-Zink-Magnesium-Legierungen. Z Metallkd. 45:577-583
Chen, K., Liu, H., Zhang, Z., Li, S., Todd, R. I., 2003. The improvement of constituent dissolution and mechanical properties of 7055 aluminum alloy by stepped heat treatments. J Mater Process Technol. 142:190-196
Dahl, O., Detert, K., 1955. Über die Aushärtung von Aluminium-Magnesium-Legierungen. Z Metallkd. 46:94-99
Hahn, G. T., Rosenfield, A. R., 1975. Metallurgical factors affecting fracture toughness of aluminum alloys. J Metall Trans A. 6:653-668
Hatch, J. M., 1984. Aluminium – Properties and Physical Metallurgy. Metals Park: ASM.
Matz, A. M., Mocker, B. S., Müller, D. W., Jost, N., Eggeler, G., 2014. Mesostructural design and manufacturing of open-pore metal foams by investment casting. Adv Mater Sci Eng. 421729;1-9
Mondal, C., Mukhopadhyay, A. K., 2005. On the nature of Ti(Al2Mg3Zn3) and S(Al2CuMg) phases present in as-cast and annealed 7055 aluminum alloy. Mat Sci Eng A. 391:367-376
Mondolfo, L. F., 1971. Aluminum alloys – structure and properties. London: Butterworths.
Müller, D. W., Matz, A. M., Jost, N., 2013. Casting open porous Ti foam suitable for medical applications. Bioinsp Biomim Nanobiomater. 2;76-83
O’Mahoney, D., Browne D. J., 2000. Use of experiment and an inverse method to study interface heat transfer during solidification in the Páramo, V., Colás, R., Velasco, E., Valtierra, S., 2000. Spheroidization of the Al-Si eutectic in a cast aluminum alloy. J Mater Eng Perform. 9:616-622
Petzow, G., 2006. Metallographisches, Keramographisches, Plastographisches Ätzen. 6th ed. Berlin: Gebrüder Bornträger Verlag.
Valdez, S., Suarez, M., Fregoso, O. A., Juárez-Islas, J. A., 2012. Microhardness, microstructure and electrochemical efficiency of an Al (Zn/xMg) alloy after thermal treatment. J Mater Sci Technol. 28:255-260
Wang, L. C., Wang, F., 2001. Preparation of the open pore aluminum foams using investment casting process. Acta Metall Sin (Engl Lett). 14:27-32
Wen, W., Zhao, Y., Morris, J. G., 2005. The effect of Mg precipitation on the mechanical properties of 5xxx aluminum alloys. Mater Sci Eng A. 392:136-144
Wunderlin, R., 1975. Beeinflussung der Warmrißneigung der Legierung G-AlZn5Mg durch Legierungszusätze. Ph.D. dissertation, TU Berlin.
Yamada, Y., Shimojima, K., Sakaguchi, M., Mabuchi, M., Nakamura, M., Asahina, T., 1999. Processing of an open-cellular AZ91 magnesium alloy with a low density of 0.05 g/cm3. J Mater Sci Lett. 18:1477-1480
Zhang, L., Eskin, D. G., Miroux, A., Subroto, T., Katgerman, L., 2012. Influence of melt feeding scheme and casting parameters during direct-chill casting on microstructure of an AA7050 billet. Metall Mater Trans B. 43:1565-1573 investment casting process. Exp Therm Fluid Sci. 22:111-12