Revealing the Early Forming Behaviors of a Carbon-Fiber-Reinforced Aluminum Foam through Synchrotron X-ray

Xi Sun, Peng Huang, Zhaoke Wang, Dongyu Fang, Qianjin Zeng and Guoyin Zu *

School of Materials Science and Engineering, Northeastern University, Shenyang 110819, China; sunxi0524@163.com (X.S.); neuhuangpeng@163.com (P.H.); zhaoke xc@163.com (Z.W.); fangdongyu11366@163.com (D.F.); cengqiangjin@163.com (Q.Z.)

* Correspondence: zugy@smm.neu.edu.cn; Tel.: +86-24-83686462; Fax: +86-24-83682912

Received: 14 November 2018; Accepted: 15 December 2018; Published: 24 December 2018

Abstract: The effect of copper-coated carbon fiber (C_{f}) on the foaming behavior of aluminum foam prepared by the powder metallurgy (PM) method, was studied, by using the synchrotron radiation technique. The corresponding stabilizing mechanism of the C_{f} was discussed and analyzed, by a comparison of the dynamic foaming process of the samples, prepared using pure Al, and that with an additional C_{f}, under the same heating regime. It was found that the C_{f}, acting as an “artificial defect” in the matrix, effectively guided the cell’s nucleation process. It not only improved the dispersion of the cell nucleation—which led to a more dispersed distribution of internal stress in the early nucleation stage—but also effectively eliminated the influence of the internal differences caused by the preparation method, which led to a more uniform distribution of cells, during the nucleation and growth stage. Thus, the cell evolution stability was greatly improved when the matrix was still in the solid phase.

Keywords: aluminum foam; powder metallurgy; synchrotron X-ray; copper-coated carbon fiber; foaming stability

1. Introduction

The nucleation, growth, stability, and rupture of cells are the main mechanisms in the formation of aluminum foam, which has to be controlled to obtain a uniform cell structure [1]. There is no doubt that the stability control during the foaming process, determines the quality of the final product, to a great extent [2]. Thus, many predecessors have devoted themselves to researching and exploring methods to improve stability, during the entire foaming process [3,4].

For the aluminum foam made by the melt-foaming method, after dispersing the foaming agent and other additives by a mechanical stirring device, the nucleation and growth of cells in the molten metal are uniformly dispersed, which creates favorable preconditions for the foaming stability, in a later period [5]. In this method, solid particles (e.g., calcium metal) are added to the molten metal, to increase the viscosity of the molten metal, adjusting its foaming properties and contributing to its foam stability [6]. Additionally, pre-oxidized foaming agent powder (e.g., titanium hydride) is used to delay the hydrogen release during the foaming process, ensuring that the temperature of the beginning of the thermal decomposition of the blowing agent is close to the liquidus temperature of the metal [7–11].

However, other methods also urgently need to be proposed, in order to strengthen the stability of the foaming process. Currently, it is generally accepted that adding substances with high melting points, into the matrix, is one simple and effective method to improve the stability of the foaming
The researchers Ip et al. [16] observed that a small amount of SiC particles is necessary to produce stable liquid-metal foam. Wang et al. proposed a stabilization mechanism [17], they suggested that the particles of SiC, could result in a decrease in the pressure difference between the plateau border and the liquid film, by modifying the curvature of the film surface, which led to more difficulty for the broken cell walls. Heim et al. [18] further explored the stabilization mechanism of SiC particles in the metal melt, and pointed out that not only are the particles required to allow for foaming, but also the formation of the oxide skin is necessary, and the combination of both is the basis of the film and foam stabilization.

Moreover, Zhuo-Kun Cao et al. [19] further proposed that a smaller amount of copper-plated carbon fibers (Cf) is better for the stability of the liquid-metal foam, and attributed to its better geometry, compared with the traditional SiC particles in the liquid–metal foam. The corresponding stable mechanism of the Cf in the liquid-metal was explained by Yongliang Mu et al. [20], who listed several specific models of the interplay process between the carbon fiber and the film.

Similar to the melt-foaming method, the foaming stability mechanism of the above-mentioned Cf is also applicable for the aluminum foam made by powder metallurgy (PM), when the matrix was in a liquid situation [21]. However, it was obviously neglected when it was below the solidus temperature, during the heating period, which seems to be equally important for the quality of the final product.

Many researchers have noticed the significance of this stage, in which the nucleation and growth of cells often take place, although the matrix was still in the solid phase [22,23]. For example, Rack et al. [24] studied the formation mechanism of micro-pores, below the melting point of the matrix, and proved that the pores prefer to form at the weakest lines in the pressed powder, rather than at the location of the individual blow agent particles. Lázaro et al. [25] indicated that the heat treatment of the aluminum foam precursors, when below a solidus temperature, can effectively reduce the initial anisotropic-semisolid expansion phenomenon, which has a detrimental effect on the whole foamability and the subsequent quality of the final cellular structure.

In this paper, the stabilization mechanism of the Cf below the liquidus temperature of the matrix, was studied by observing the dynamic foaming process, by using the Synchrotron radiation technology [11,26–30]. The corresponding data regarding the evolution of cells at the early nucleation and growth stages, were gained. Through a comparison of the evolution of cells, within the pure Al sample, under the same heating regime, the beneficial effects of the Cf on improving foaming stabilization when the matrix was still in the solid phase, were determined, which provided a necessary theoretical basis for the comprehensive application of carbon fiber in the preparation of aluminum foam, by PM [31].

2. Experimental Procedures

2.1. Sample Preparation

Foamable precursors were manufactured by compacting mixtures of aluminum (Al), titanium hydride (TiH₂), and magnesium (Mg) powder. The Al powder that was used had a nominal purity of 99.8% (metallic impurities) and a mean diameter of 48 mm, as specified by the manufacturer (Hong Yu, Beijing, China). Pre-oxidation treatment on the TiH₂ powder, with a mean diameter of less than 5 µm, was conducted in air, at 450 °C, for one hour [7,8,10,11]. The Mg powder (purity 99.8%, nominally diameter <48 µm) and the Cf (<3 mm in length, the copper layer on the surface with a thickness of 1.4 µm, obtained by an electroplating method) were applied. Al powder was mixed with 0.5 wt.% TiH₂ and 0.5 wt.% Mg powder in a three-dimensional mixing machine for 180 min, in group one, while 0.5 wt.% Cf acted as another added substance, in group two. The mixtures were subjected to uniaxial cold pressing for 10 min, in a cylindrical mold, under a pressure of 400 MPa, and the pressing was carried out in air. Cylindrical tablets (50 mm diameter, ~15 mm high) were obtained. Samples were cut out of the pressed tablets by electric spark cutting, with dimensions of 10 × 10 × 3 mm³; then, they were polished through a manual operation, to ensure a flat surface perpendicular to the
thickness direction. A sample thickness of less than two mm is required for the synchrotron radiation because of the low light intensity of the 4W1A (Institute of High Energy Physics, Chinese Academy of Sciences, Beijing, China), which was the experimental station of the Beijing Synchrotron Radiation Research Institute.

2.2. Synchrotron Radiation Experiments

Foaming was carried out in an X-ray transparent, semi-gas tight crucible, as shown in Figure 1. The crucible consisted of a certain number of ceramic sheets (50 mm long, 25 mm wide, and 0.38 mm wall thinness) seated in a miniature heating furnace (Homemade devices, Shenyang, China), with a hole across it (15 mm long and 15 mm wide). The ceramic sheet at both ends was solid, while the ceramic sheets inside it had rectangular holes (15 mm long and 15 mm wide). The thinner nonporous ceramic sheet on both sides made it easier for X-rays to pass through, while the number of porous ceramic sheets inside, ensured a sufficient expansion space for the sample and made it convenient for the internally generated gas to be removed through the gap. The semi-gas tight crucible where the samples were foamed, had a maximum heating power of 2 KW, which was generated by using the furnace outside. The sample temperature was measured by two thermocouples, which were led through the heating furnace and placed very close to the crucible, on both sides of the sample.

![Figure 1. Sketch of the foaming furnace.](image)

Two Proportion Integration Differentiation (PID) controllers allowed for a predefined temperature course, which included heating up for 10 min at a designated heating power, leading to a heating rate of, approximately, 50 K·min\(^{-1}\), for the sample size used, approaching the setting temperature of 550 °C, and then slowing down the heating rate by, approximately, 5 K·min\(^{-1}\), in order for it to be easily recorded by a charge-coupled device (CCD, VHR-16M, Photonic Science and Engineering Limited, Saint Leonards-on-sea, UK.) within the X-ray station, until a final temperature of 690 °C, and then holding at this temperature for one further minute.

The device for real-time, in situ, X-ray radioscopy, during sample foaming, was based on the 4W1A experimental station in the Beijing Synchrotron Radiation Facility (BSRF, Institute of High Energy Physics, Chinese Academy of Sciences, Beijing, China). The intensity of the selected X-ray source was 20 KeV, which represents the maximum ability of the Synchrotron Radiation Facility to ensure the maximum penetration efficiency accepted by the CCD. The maximum view size (7.4 mm × 7.4 mm) was used, which was convenient to observe the whole foaming process.

An exposure time ranging from 300 ms to one second, varied according to the thickness and composition of the sample, and this led to different light transmittance values being selected to ensure that a certain photoelectron intensity was received by the CCD, in one picture. It is worth noting
that the photoelectron intensity should neither be too high nor be too week. A suitable photoelectron intensity should be received by the CCD device, to form a clear image, without causing harm to it.

The synchrotron radiation image was subjected to image processing, before the collection of cell evolution data. The Quick HDR software (Version 1.0, MediaChance Limited, Ottawa, Canada) was used for equalizing the macroscopic brightness, to eliminate the influence of the brightness gradient, making the threshold easier for maximizing the characterization of the cells from the image. After adjusting the contrast, brightness, background elimination, and the corresponding Gaussian fuzzy algorithm function, ImageJ software (Version 1.49, National Institutes of Health, Bethesda, MD, USA) was used to achieve more clearly binarized images [32]. After that, image data that was batch converted out by the Image Pro-Plus software, was finally shown out by curves, using the Origin 9.0 software (Version 9.0, OriginLab Corporation, Northampton, MA, USA).

2.3. Sample Characterization

For the foam precursors, electro-discharge machining was applied to the section, and grinding papers of 1000–3000 grit, were used for mechanical grinding, which was then followed by a manual final polishing with an oxide suspension (MgO) on deerskin. A ZEISS scanning electron microscope (SEM, ZEISS, Jena, Germany) was used to observe the coated carbon fiber that settled on the surface, after careful polishing. In order to enhance the visibility of the copper, alcohol was introduced to prevent the oxidation of the copper when being polished.

3. Results and Discussion

Figure 2 shows the typical foam evolution sequences of aluminum foam that had 0.5 wt.% C\textsubscript{f} added during the early stage, which was detected by the synchrotron radiation. It could be seen that the C\textsubscript{f} was evenly distributed in the precursor when the temperature was below 580 °C, as shown in Figure 2a. However, the C\textsubscript{f} (existing in Figure 2a) gradually became harder to distinguish in Figure 2b, when the temperature reached 590 °C. Moreover, until the temperature reached 610 °C, the C\textsubscript{f} eventually disappeared in Figure 2c, which was accompanied by the formation of large numbers of micro-pores, which preferred to nucleate around the C\textsubscript{f} at that time. There were no obvious changes that took place for the micro-pores, until 652 °C, as shown in Figure 2d, which we defined as the demarcation point of the initial nucleation stage and the nucleation and growth stage. During the subsequent nucleation and growth stage, when the temperature was between 652 °C and 660 °C, the cells began to nucleate and grow up rapidly, and the dynamic foaming process in this stage were shown in Figure 2d,e. The growth and coalescence stage of the cells took place above 660 °C (see Figure 2e,f), at which the cells grew even more quickly accompanied with the large scale of cell coalescence, at the same heating rate. At this stage, the cells that were rich in C\textsubscript{f} prior to growth, and so as had easier contact with each other, created more favorable condition for the cells’ coalescence nearby, which could be seen through the dynamic imaging process.

Figure 3 shows the foam evolution sequences for the sample made of pure Al, detected by synchrotron radiation. It can be seen that no visible micro-pores formed in Figure 3a when the temperature was below 607 °C, however, the decomposition temperature had been already reached for the titanium hydride [9,11]. Nucleation seems to be more difficult for the pure Al, compared with the sample with an additional 0.5 wt.% C\textsubscript{f}, at a similar heating temperature as that shown in Figure 2c, in which a large number of micro-pores had already been clearly shown. Figure 3b,c show the cell evolution with the temperature rising from 627 °C to 630 °C. During this heating process, the nucleation rate of the cells obviously accelerated, compared with the process of the previous temperature interval. There were no obvious changes that took place again in the cell nucleation, after that, until a temperature of 652 °C was reached, which is shown in Figure 3d. Compared with the sample with an additional C\textsubscript{f}, shown in Figure 2d, under the same heating condition, relatively less quantitation but bigger sizes of cell nuclei was found. During the defined sequential nucleation and growth stage (see Figure 3d–h), the corresponding temperature range of which was between 652 °C
and 660 °C, the foaming process became quicker, even though a heating rate lower than 5 K-min\(^{-1}\) was adopted. Cells in the aggregation (region A in Figure 3e–g) grew more rapidly for the moment and was easier for the merge and collapse, which contributed to the more extremely uneven cell distribution, compared to the foaming process of the aluminum foam, which had an additional C\(_f\) inside (see Figure 2d,e). Figure 3h,i show the typical morphology for the foaming process of pure Al, when the temperature was above 660 °C. Cells grew and merged in large scales and were distributed extremely non-homogeneously. Small cells that formed later, were more likely to move towards the region where the cells were in aggregation (shown in region A), and eventually formed an abnormal growth, which resulted in a more serious uneven cell distribution (see Figure 3i, at the temperature of 663 °C).

![Figure 2](image-url)

**Figure 2.** Foaming process of the aluminum foam with additional 0.5 wt.% C\(_f\): (a) <580 °C; (b) 590 °C; (c) 610 °C; (d) 652 °C; (e) 660 °C; and (f) 663 °C.

It is generally accepted that the different elements of light absorption were different, according to the principle of the synchrotron radiation imaging. For the samples that were passed through an X-ray, across the same thickness, a more clear contrast of image was gained when there was a greater difference between the element atomic numbers [27]. However, owing to the small difference in the atomic numbers, between the carbon fibers and the aluminum-based elements, it was hard to observe the carbon fiber in the field-view, due to limited equipment capacity. Coincidentally, the element of copper which has a bigger difference in atomic number, when compared with the aluminum matrix, was planted on the surface of the carbon fiber, in this experiment, making the distribution of the carbon fiber more clearly seen in the field-view when the temperature went below 580 °C, just as shown in Figure 2a. However, the image of the carbon fibers gradually became obscure and disappeared evenly, when the temperature rose to 590 °C, as shown in Figure 2b. In order to understand the evolution mechanism of the C\(_f\), in-depth, SEM was used to detect the situation of the C\(_f\), in the aluminum foam, with different heat treatment temperatures, as showed in Figure 4. No heat treatment was applied to Sample (a), while a 15 min heat treatment was applied to both Sample (b) and Sample (c), at different temperatures of 570 °C and 600 °C, respectively. It can be seen that the C\(_f\) was heavily copper-coated,
in Sample (a), while the copper layer was partially reserved on the carbon fiber surface, for Sample (b). For Sample (c), no copper layer was detected on it.

Figure 3. Foaming process of aluminum foam made of pure Al: (a) <607 °C; (b) 627 °C; (c) 630 °C; (d) 652 °C; (e) 654 °C; (f) 656 °C; (g) 658 °C; (h) 660 °C; and (i) 663 °C.

However, in the image sequences of Figure 2, the C_f began to disappear when the temperature was more than 580 °C. While compared with Figure 4b, the copper layer had already dissolved when the temperature was no more than 570 °C. It could be attributed to a longer heating time for the sample shown in Figure 4b, in which the copper clad layer had enough holding time to diffuse (570 °C, 15 min), while as a comparison, for the sample shown in Figure 2, there was nearly no holding time at all, leading to the disappearance of the copper layer, at temperatures higher than 580 °C. Thus, it could be concluded that the disappearance of the C_f shown in the synchrotron radiation field of Figure 2 was closely related to the dissolution of the copper layer on the surface of the carbon fibers. Additionally, with increasing temperatures, accompanied by a large number of cell nuclei that were generated and attached to the surface of the C_f, the black carbon fiber in the original imaging field of Figure 2a, completely turned white, as shown in Figure 2c. Thus, the C_f that existed in the matrix
of the aluminum foam effectively promoted the PM-guided nucleation process, during the early nucleation stage.

![SEM micrograph of the Cf with heat treatment](image)

**Figure 4.** SEM micrograph of the Cf with heat treatment of (a) atmospheric temperature; (b) 570 °C, 15 min; and (c) 600 °C, 15 min.

Figure 5 shows the curve of the cell area (sum and average) of the foamed aluminum samples (with and without Cf) in the early nucleation stage (<652 °C), according to the statistics of Figures 2a–d and 3a–d. Both samples were in the same size and adopted the same heating process. As shown in Figure 5a, for the foamed aluminum sample with an additional Cf inside, the initial nucleation temperature was lower than 590 °C, and the cell area (sum) showed a slow linear growth, with rising temperatures. As a comparison, for the aluminum foam made of pure Al, no obvious nucleation occurred when the temperature was below 607 °C, the nucleation temperature occurred mainly between 610 °C and 652 °C. The curve changed, slowly, with a rising temperature, in all of the early nucleation stage, except for the significant turning point of about 630 °C (the main nucleation temperature), at which the cell nuclei formed quickly, to a certain size, while no obvious change took place in them, until after 652 °C. Under the same heating conditions, the cell area (sum) of the foamed aluminum-containing Cf was higher than that of the pure Al one.

Figure 5b shows the curve of the cell area (average) in the early cell nucleation stage, for both samples, with and without any Cf inside. Similar to the curve (sum) of the aluminum foam made by the pure Al, nearly no obvious change took place below 607 °C, because of a low cell nucleation rate. When the temperature reached up to 610 °C, the cell area (average) increased rapidly and quickly exceed that of the Cf, while no obvious change occurred for the cell area (sum), due to the low rate of cell nucleation, in pure Al. The nucleation rate slowed down in the rest of the early cell nucleation stages, but had a turning point at about 630 °C, at which, the average cell area increased rapidly, while no obvious change took place for cell area (average) of the Cf. Under the same heating conditions, the curve of the cell area (average) of the sample made of pure Al, was more unstable, which showed obvious ladder-like growth characteristics, compared to the sample with an additional Cf, which had a curve that was nearly unchanged, in all early cell nucleation stage.
Obviously, by an addition of Cf to the matrix, the dispersion of the cell nuclei increased, greatly, which resulted in different distributions of the two curves. Due to the high strength and poor plasticity of the matrix, owing to low temperatures (more than 10 °C lower than the liquid–solidus temperature), deformation in the foaming process became more difficult in all early nucleation stages. However, with temperatures rising and the heating time extending, more hydrogen accumulated continuously, which contributed to a higher internal pressure due to the inevitably decomposition of the titanium hydride. Under the same conditions (same content of the titanium hydride and the same heating system), although there was a larger nucleation area (sum) for the sample with an additional Cf inside, the cell area (average) was relatively smaller, owing to larger numbers of higher dispersion of cell nuclei. Therefore, the internal pressure that was made by the decomposition of the titanium hydride in the matrix, was more dispersed, thus, stress concentration was effectively avoided, which of course, led to a higher substrate stability. As a comparison, for the aluminum foam made of pure Al, cell nucleation became even more difficult for the lack of Cf, which effectively acted as an “artificial defect” ingredient in the matrix-guided cell nucleate process. When the nucleation ability that was caused by the decomposition of the titanium hydride overcame the matrix restrictions and began to nucleate, with fewer cell nucleation particles and a much higher temperature, internal pressure inevitably accumulated to a much higher point than that of the Cf. Thus, the nucleation process suddenly accelerated, at a certain point, accompanied with the formation of bigger cell nuclei, which could be directly observed through the dynamic imaging shown in Figure 3 (also can be analyzed by the curves of Figure 5, at temperatures of 610 °C and 630 °C, respectively). As the internal pressure decreased sharply with a rapid increase of the average cell area, afterward, the nucleation process slowed down again. Only when the internal pressure reached the matrix’s deformation resistance limitation, again, the cell nuclei grew rapidly, once, which contributed to a ladder-like growth curve of the average cell area.

Thus, through an addition of Cf, a more stable situation of the cell nucleation process was gained, in all early nucleation stage. Local stress concentration, caused by hydrogen accumulation was effectively avoided for the scattering of the cell nuclei, which in turn, effectively increased the matrix’s maximum capacity for hydrogen, in the same external environment.

Figure 6 shows the curves of the cell area (sum and average) of the foamed aluminum samples (with and without Cf), during the nucleation and growth stage (652–660 °C), and the corresponding real-time excerpt images are shown in Figure 2d,e and Figure 3d–h. At this stage, the cell evolution significantly accelerated, compared to the early nucleation stage before.

Similar to the foam process in the early nucleation stage, for the foam sample with an additional Cf inside, a smoother and more stable increase of the total cell area was found (shown in Figure 6a), compared with that of the pure Al. At temperatures of 655 °C, the cell area (sum) for the pure Al had

![Figure 5](image-url)
already exceeded that of the sample with an additional Cf, and increased faster and faster, with the rising temperature.

Figure 6. The curve of the cell area of the aluminum foam in the nucleation and growth stage: (a) the sum cell area; (b) the average cell area.

Figure 6b shows the curves of the average cell area of these two samples. It can be seen that the area showed an obvious increase for the pure Al, while only a limited lift was gained for the sample with the additional Cf inside. It could be better understood by a comparison of the dynamic foaming process of these two samples, in the image sequences of Figure 2d,e and Figure 3d–h. That is, for the sample with the Cf, the cells showed multitudinous and uniform distribution in the sequences of Figure 2d,e, while for the sample made of pure Al, the cells preferred to be concentrated in some regions (see region A), improving the probability for the cells to contact and merge with each other. Eventually, the value of the average cell area for the pure Al increased much faster than the sample with an additional Cf, as shown in Figure 6b.

Different from the aluminum foam made by the melt foaming method, in which the cell evolution only takes place above the liquid temperature, the nucleation process had occurred when the temperature was lower than 610 °C in the PM. Being in a solid state with a high-strength matrix, the regions with a lower deformation resistance preferred to nucleate and grow with the continued decomposition of titanium hydride. The uniform distribution for cell evolution in the nucleation and growth stage is extremely important, for lots of obvious defects formed later, usually closely related to the uneven distribution, at this stage.

However, the ideal sample was hard to be prepared in practical experiments, that is, varied internal differences unavoidably existed in the matrix, which led to different deformation resistance, in different regions. The internal differences were non-homogeneously distributed in the matrix. To make matters worse, sometimes even gradient-distributed (for example, the manufacturing process of the uniaxial compression of the die extrusion, might lead to uneven internal stress distribution along the radius direction in the cylindrical samples) [26]. Owing to these internal differences, the critical temperature of the nucleation and growth of cells, in the matrix, were different at different sections, which of course led to non-uniform nucleation and growth distribution, at different sections of the sample, for the same heating temperature. Thus, just as the foam sample of pure Al (shown in Figure 3d–h), non-uniform distribution for cell nucleation and growth was gained, when the temperature rose above 652 °C. Cells preferred to nucleate and grow at Section A (see Figure 3f), where there was a poor matrix strength, while no visible nucleation process even took place for the other section where there was a stronger deformation resistance. With the foaming process going on, the merge process of the cells in Region A, were unavoidable. Thus, deformation resistance in this region was further weakened, which resulted in a sharper uniformity decrease for the cell distribution, when the temperature was no more than 660 °C.

Figure 7 shows the schematic diagram of the mechanism for defects formation, which is more likely to occur in pure Al samples, caused by the cell concentrations to be distributed during the nucleation
and growth stage. Region A represents the areas where the cells showed a high concentration. With the foaming process going on, high internal stress accumulating constantly, resulted in a deformation in the nuclear-intensive region, to a certain extent. However, due to the restrictions of the surrounding matrix, the shear stress between the nuclear-intensive and less nucleated region became bigger and bigger, with the deformation increasing. As the shear stress caused by the hydrogen release reached the yield limit of the matrix material, the matrix broke down and cracks formed easily, when it was under liquids temperature. Thus, a large through-hole formed in the shear band, which contributed to an abnormal cell growth, afterward. With the heating process going on, lots of small cells that formed in the less nucleated region, tended to move to Region A that resulted in an even worse cell distribution being generated, which has not been shown in this article.

![Figure 7. The schematic for the formation of shear stress.](image)

Similar to the sample made of pure Al, many diversities also exist in the sample with an additional Cf inside. However, the influence of diversities seems to have been effectively reduced. With a lower bonding strength and narrow gaps connected in the matrix, the Cf was considered to be “artificial defects” particulates in the matrix, creating more favorable conditions for the cell nucleation and growth process. These “artificial defects” with a stronger guiding ability for the cell nucleation and growth than the internal defects, were distributed in the matrix, evenly, and were more scattered. Thus, the characteristics of the gradient distribution for the cell nucleation and growth of the sample made of pure Al, was effectively eliminated by the addition of the Cf, leading to a more uniform cell nuclei distribution for all cell nucleation and growth stage, macroscopically.

In summary, for aluminum foam made of PM, Cf addition, as an effective stabilizing material, played an important role in stabilizing the cell evolution process of the foamed aluminum, in the early stage. Although large-scale coalescence occurred in the later stages, which might have led to certain cell non-homogeneity [26], the macroscopic distribution of the cell evolution had already been determined by the cell evolution at the nucleation and the growth stage, to a great extent, already. Thus, controlling the uniformity of the cell distribution at the nucleation and growth stage, was an essential precondition, which should be paid more attention for the improvement of the cell quality for aluminum foam made of PM.

4. Conclusions

1. The copper-coated carbon fiber in the aluminum foam matrix could effectively guide the nucleation process. The nucleation temperature was reduced and the nucleation process was more conducive.
The stability of the cell’s nucleation process, during the early cell nucleation stage ($\leq 652 ^\circ C$), was reinforced by the addition of $C_f$ inside the matrix. More cell nucleation dispersion in the matrix contributed to a higher stress dispersion, thus, a higher deformation resistance was gained for the matrix, which resulted in a nucleation process that was more gradual and easier to control.

The influence of the non-homogeneously distributed internal differences were significantly eliminated by the addition of $C_f$ to the matrix, during the nucleation and growth stage (652–660 $^\circ C$). The cells’ preferential growth and merger of the region which was poor in deformation resistance, caused by a non-homogeneous distribution of internal differences, was effectively eliminated by the notable guiding ability of the $C_f$, in the matrix. Thus, fewer defects formed when the matrix was still in the solid phase, which created a more favorable foaming environment, afterward.

Author Contributions: X.S. and P.H. conceived and designed the experiments; X.S. and P.H. performed the experiments; X.S., Z.W., D.F. and Q.Z. analyzed the data; X.S. wrote the paper; G.Z. provided the research ideas, led the relevant research design, guided the manuscript revision, created the necessary conditions for the research, finalized the version to be published.

Funding: This work was funded by the National Natural Science Foundation of China (Grant Number U1332110) and the Fundamental Research Funds for the Central Universities of China (Grant Number N170212022).

Conflicts of Interest: The authors declare no conflict of interest.

References
1. Banhart, J. Manufacture, characterisation and application of cellular metals and metal foams. *Prog. Mater. Sci.* 2001, 46, 559–632. [CrossRef]
2. Banhart, J. Metal foams: Production and stability. *Adv. Eng. Mater.* 2006, 8, 781–794. [CrossRef]
3. Asavavisithchai, S.; Kennedy, A.R. The effect of Mg addition on the stability of Al-Al2O3 foams made by a powder metallurgy route. *Scr. Mater.* 2006, 54, 1331–1334. [CrossRef]
4. Korner, C.; Arnold, M.; Singer, R.F. Metal foam stabilization by oxide network particles. *Mat. Sci. Eng. A* 2005, 396, 28–40. [CrossRef]
5. Garcia-Moreno, F. Commercial Applications of Metal Foams: Their Properties and Production. *Materials* 2016, 9, 85. [CrossRef] [PubMed]
6. Yang, C.C.; Nakae, H. The effects of viscosity and cooling conditions on the foamability of aluminum alloy. *J. Mater. Process. Technol.* 2003, 141, 202–206. [CrossRef]
7. Matijasevic, B.; Banhart, J. Improvement of aluminium foam technology by tailoring of blowing agent. *Scr. Mater.* 2006, 54, 503–508. [CrossRef]
8. Matijasevic-Lux, B.; Banhart, J.; Fiechter, S.; Gorke, O.; Wanderka, N. Modification of titanium hydride for improved aluminium foam manufacture. *Acta Mater.* 2006, 54, 1887–1900. [CrossRef]
9. Yang, D.H.; Chen, J.Q.; Wang, H.; Jiang, J.H.; Ma, A.B.; Lu, Z.P. Effect of Decomposition Kinetics of Titanium Hydride on the Al Alloy Melt Foaming Process. *J. Mater. Sci. Technol.* 2015, 31, 361–368. [CrossRef]
10. Kennedy, A.R.; Lopez, V.H. The decomposition behavior of as-received and oxidized TiH2 foaming-agent powder. *Mater. Sci. Eng. A* 2003, 357, 258–263. [CrossRef]
11. Jimenez, C.; Garcia-Moreno, F.; Pfretzschner, B.; Klaus, M.; Wollgarten, M.; Zizak, I.; Schumacher, G.; Tovar, M.; Banhart, J. Decomposition of TiH2 studied in situ by synchrotron X-ray and neutron diffraction. *Acta Mater.* 2011, 59, 6318–6330. [CrossRef]
12. Duarte, I.; Ferreira, J.M.F. Composite and Nanocomposite Metal Foams. *Materials* 2016, 9, 79. [CrossRef] [PubMed]
13. Kumar, N.V.R.; Rao, N.R.; Sudhakar, B.; Gokhale, A.A. Foaming experiments on LM25 alloy reinforced with SiC particulates. *Mater. Sci. Eng. A* 2010, 527, 6082–6090. [CrossRef]
14. Heim, K.; Vinod-Kumar, G.S.; Garcia-Moreno, F.; Banhart, J. Stability of various particle-stabilised aluminium alloys foams made by gas injection. *J. Mater. Sci.* 2017, 52, 6401–6414. [CrossRef]
15. Afshar, M.; Mirbagheri, M.H.; Movahedi, N. Effect of SiC particle size on the mechanical properties of closed aluminum foams. *Mater. Test.* 2017, 59, 571–574. [CrossRef]
16. Ip, S.W.; Wang, Y.; Toguri, J.M. Aluminum foam stabilization by solid particles. *Can. Metall. Q.* 1999, 38, 81–92. [CrossRef]

17. Wang, D.Q.; Shi, Z.Y. Effect of ceramic particles on cell size and wall thickness of aluminum foam. *Mater. Sci. Eng. A* 2003, 367, 45–49.

18. Heim, K.; Vinod-Kumar, G.S.; Garcia-Moreno, F.; Rack, A.; Banhart, J. Stabilisation of aluminium foams and films by the joint action of dispersed particles and oxide films. *Acta Mater.* 2015, 99, 313–324. [CrossRef]

19. Cao, Z.K.; Li, B.; Yao, G.C.; Wang, Y. Fabrication of aluminum foam stabilized by copper-coated carbon fibers. *Mater. Sci. Eng. A* 2008, 486, 350–356. [CrossRef]

20. Mu, Y.L.; Zu, G.Y.; Cao, Z.K.; Yao, G.C.; Wang, Q.D. Metal foam stabilization by copper-coated carbon fibers. *Scr. Mater.* 2013, 68, 459–462. [CrossRef]

21. Liang, L.S.; Wang, L.; Luo, S.; Peng, J.X.; Ma, N.N.; Wang, G.H. Effect of Short Copper-Coated Carbon Fibers on Stabilization of Al Foams Made by Powder Metallurgy Route. *Nanosci. Nanotechnol. Lett.* 2017, 9, 1415–1418. [CrossRef]

22. Dudka, A.; Garcia-Moreno, F.; Wanderka, N.; Banhart, J. Structure and distribution of oxides in aluminium foam. *Acta Mater.* 2008, 56, 3990–4001. [CrossRef]

23. Ding, X.; Liu, Y.; Chen, X.; Zhang, H.W.; Li, Y.X. Optimization of cellular structure of aluminum foams produced by powder metallurgy method. *Mater. Lett.* 2018, 216, 38–41. [CrossRef]

24. Rack, A.; Helwig, H.M.; Butow, A.; Rueda, A.; Matijasevic-Lux, B.; Helfen, L.; Goebbels, J.; Banhart, J. Early pore formation in aluminium foams studied by synchrotron-based microtomography and 3-D image analysis. *Acta Mater.* 2009, 57, 4809–4821. [CrossRef]

25. Lazaro, J.; Solorzano, E.; Rodriguez-Ferez, M.A.; Ramer, O.; Garcia-Moreno, F.; Banhart, J. Heat treatment of aluminium foam precursors: effects on foam expansion and final cellular structure. *Procedia Mater. Sci.* 2014, 4, 287–292. [CrossRef]

26. Mukherjee, M.; Garcia-Moreno, F.; Banhart, J. Defect generation during solidification of aluminium foams. *Scr. Mater.* 2010, 63, 235–238. [CrossRef]

27. Tolnai, D.; Townsend, P.; Requena, G.; Salvo, L.; Lendvai, J.; Degischer, H.P. In situ synchrotron tomographic investigation of the solidification of an AlMg4.7Si8 alloy. *Acta Mater.* 2012, 60, 2568–2577. [CrossRef]

28. Garcia-Moreno, F.; Babcsan, N.; Banhart, J. X-ray radioscopy of liquid metal foams: influence of heating profile, atmosphere and pressure. *Colloids Surf. A* 2005, 263, 290–294. [CrossRef]

29. Zu, G.Y.; Sun, X.; Huang, P.; Sun, S.L. Study on Powder Metallurgical Foaming Mechanism of Aluminum Foam Sandwich Using SR-CT. *Rare Metal Mater. Eng.* 2017, 46, 3141–3145.

30. Duarte, I.; Oliveira, M.; Garcia-Moreno, F.; Mukherjee, M.; Banhart, J. Foaming of AA 6061 using multiple pieces of foamyable precursor. *Colloid Surf. A* 2013, 438, 47–55. [CrossRef]

31. Aggogeri, F.; Borboni, A.; Merlo, A.; Pellegrini, N.; Ricatto, R. Vibration Damping Analysis of Lightweight Structures in Machine Tools. *Materials* 2017, 10, 297. [CrossRef] [PubMed]

32. Duarte, I.; Ferreira, J.M.F. 2D Quantitative Analysis of Metal Foaming Kinetics by Hot-Stage Microscopy. *Adv. Eng. Mater.* 2014, 16, 33–39. [CrossRef]