Effects of CuSn33 content on the microstructure and mechanical properties of Al/Cu bimetallic foams

Qiang Feng1, Shengshou Ma2, Changzhong Liao*, Yuhuan Xie1 and Zhiwei Duan1

1 Institute of Nonferrous Metal Gradient Materials Preparation and Performance, College of Physics and Engineering, Chengdu Normal University, No.99 Haike Road, Eastern Section, Chengdu, Sichuan 611130, People’s Republic of China
2 Department of Civil Engineering, The University of Hong Kong, Pokfulam Road, Hong Kong 999077, People’s Republic of China

E-mail: liaocz29@connect.hku.hk

Keywords: bimetallic foams, mechanical property, powder technology, microstructure

Abstract

A new type of Al/Cu bimetallic foams with double grade plateau stresses was prepared by employing powder metallurgy sintering method. The addition of CuSn33 is to improve the microstructure and mechanical properties of Al/Cu bimetallic foams. The effects of CuSn33 content on the phase components, lattice parameter, crystallize size and microstrain of Cu-matrix were investigated with utilizing Rietveld method. Results show that there is a linear relationship between CuSn33 content and lattice parameter, and the linear fitting equation was also simulated. The addition of CuSn33 can refine the crystal lattice and increase the microstrain when the Sn with larger atomic radius dissolves into Cu-matrix alloy. However, the microstrain value drops dramatically when the precipitated δ-phase appears from the supersaturated solute elements. Segregation caused by the residual Sn in the Cu-matrix primarily appears near to the border of Al2Cu and Cu-matrix. The compressive stress-strain curves of Al/Cu bimetallic foams with 5 typical stages are radically different from the single component metal foams, including twice quasi-linear regions, twice collapse plateaus, and the densification region. The effects of CuSn33 content on the elastic modulus of Al/Cu bimetallic foams, microhardness of Cu-matrix and the absorbed energy density have also been investigated.

1. Introduction

Al foams have been attracted considerable attention because of their peculiar mechanical, acoustic, thermal, electrical and chemical properties [1–3]. Structurally, a weighty consideration is the energy absorption capacity in connection with porosity, collapse plateau stress and ductility in compression. Functionally, such as thermal insulation, sound absorption, heat dissipation and catalyst support for the open-cell are particularly important. The carbamides as spherical space holder to prepare open-cell metal foams has been paid close attention all along. Although the pores are spherical shape, the manufactured Al foams still leave much to be desired due to the low mechanical properties [3, 4]. To achieve better properties, thickening the struts was used for enhancing Al foams [4–6]. However, these approaches result in a decrease in foam porosities. Therefore, to improve the mechanical property of single Al-matrix foams via thickening the struts is likely incompatible.

Bimetallic materials combine the advanced characteristics of two metals in a single product. Several bimetallic materials, such as the Al and Cu or Cu and steel have been intensively studied [7–13]. With the development of manufacture technology, Al/Cu bimetallic materials have got an extensive application. According to the practical applications, bimetallic with various geometries can be obtained as roll bonded Al/Cu metal laminated material [7], asymmetrical roll bonding productions [8], axisymmetric extrusion manufacture [9–11] and friction welded Al/Cu bimetallic joints [12, 13]. Bimetallic Cu-Clad Al wires possess the advantages of the low Al weight as well as high conductivity and low corrosion of Cu, which is obtained by thermomechanical cladding between Al core and Cu sleeve [14]. The mechanical properties and diffusion behavior of the Al/Cu bimetallic composite have been investigated [7, 15–19]. The study shows that the interfacial strength depends on the generated phases in the interfacial region during the roll-bonded Al/Cu
laminates [7]. Some researchers pointed out that Al_{13}Cu was the first phase relative to the AlCu and Al_{4}Cu_{9} when the Al/Cu laminates prepared by plasma activated sintering at the temperature between 400 °C and 500 °C [17]. Simultaneously, several welding technologies have been used to manufacture Al/Cu laminate composites such as friction welding [13], roll bonding [18], and reactive diffusion bonding [19]. All of these processes are carried out below the eutectic temperature (548.2 °C), which demonstrates that the combination of pressure and heat is extremely important to the solid-state bonding between the two different metals [15, 18]. It is noted that the Al/Cu bonding is dissimilar when the sintering temperature range between eutectic temperature and Al melting point, and the evolutionary process of intermetallic phase formation at the Al/Cu interface is uncertain [20]. The interfacial intermetallic phases at the Al/Cu interface were also investigated when the isothermal heating at 550 °C, and the related mechanism of phase formation was discussed on the basis of the experimental results [21]. However, few studies have investigated the association of Al/Cu two foam metals together to form a single composite, which makes it possible to combine the high strength of Cu and lightweight of Al.

Powder metallurgy route is an efficient control method for the cell shape, cell size and porosity. The melting point of Cu is much higher than that of Al in the bimetallic compactor. When the sintering temperature is too high, the properties of the Al-matrix will be reduced and even generate the defects of over-burning [2]. On the contrary, the sintering performance of the Cu-matrix will decrease as the temperature is too low. Therefore, it is important to choose the proper chemical constituents to adjust the temperature during the sintering processing [22]. To optimize the sintering effect of Cu-matrix, a material with a low melting point, coupled with high solubility in the Cu can be filled in the Cu powder particles. It can be obtained from the binary phase diagrams that the Sn possesses lower melting point and solubility limits in Cu [23, 24]. When the Sn atoms across the entire Cu-matrix, the solidus temperature approach 8006 °C, which may reduce the difference of sintering temperature between Al-matrix and Cu-matrix alloy [22]. However, it is liable to generate solute segregation and microstructure segregation in Cu-matrix due to the metal Sn with a low melting point. This most likely to produce a number of intermediate phases between both metals, resulting in a reduction of mechanical property of Cu-matrix [22]. Thus, it is especially important to avoid the hazards of segregation in Cu-matrix by improving the melt point of filler material. In this study, combining the high strength of Cu and the weightlessness of Al to acquire a new type of Al/Cu bimetallic foams structural material will be carried out. The pre-alloyed powder CuSn_{33} as filler in Cu-matrix to improve the microstructural and mechanical property was investigated.

2. Experimental details

2.1. Raw materials

The purities of Cu powder (Shijiazhuang Jingyuan Powder Material Co. LTD) and Al powder (Xingtai Chuangying Metal Materials Co. LTD) used in this work are ≥99.8%, and the powder particle sizes are approximately 1 μm and 15 μm, respectively. The CuSn_{33} alloy powder (Shijiazhuang Jingyuan Powder Material Co. LTD) with 33.0 wt% Sn content was prepared via the water atomization method, with an average particle size of 15 μm, and the purity is ≥99.0%, which was chosen as the filler material in Cu-matrix. The carbamide particles with diameter ranged from 0.6 to 1.8 mm were used as space holder in the present work.

2.2. Sample preparation

The preparation processes of the Al/Cu bimetallic foams are schematically shown in figure 1. Firstly, the Cu-matrix powders included Cu and CuSn_{33} (CuSn_{33} powders in the mixture are 0 wt%, 5.0 wt%, 10.0 wt%, 20.0 wt%, 30.0 wt%, namely, 1#, 2#, 3#, 4# and 5#, respectively) were homogeneously mixed in a grinder made by our laboratory. Aiming to improve the bonding in green powder compact, the aluminum dihydrogen phosphate as the adhesives were added in the powder Cu-matrix and Al-matrix are 8.0 vol% and 4.0 vol%, respectively. The carbamide particles were added evenly in the prepared mixed Cu-matrix powders, and then the mixture was filled in the mould cavity for Cu-matrix green by pressing gently. Secondly, Al powders were mixed with aluminum dihydrogen phosphate powder, and carbamide particles were filled in the previous mould cavity to get Al-matrix green after the punch was pulled out. Thirdly, the ready-pressed powders were uniaxially cold compacted in a cylindrical die cavity with a hydraulic press (Pressure 300 MPa, 15 min). The diameter and the height for the prepared cylindrical green compacts are 20 mm and 22 mm, respectively. The height of the represented Al/Cu bimetallic foams is evenly split between the Cu-matrix and Al-matrix. Lastly, the compacted pellets were dried at 200 °C in an evacuated quartz tube for 1 h, followed by heating at 610 ± 5 °C for 90 min before being cooled to the room temperature in vacuum quartz tube. The porosities of Al/Cu bimetallic foams depend on principally the carbamide particles volume according to the densities of components are 48 vol% and 48 vol% in Al-matrix and Cu-matrix, respectively.
2.3. Characterization
The XRD data (2-Theta scan range: 10–80 Deg.) of the flat and smooth cut surfaces for both Cu-matrix and interface of Al/Cu bimetallic foams were employed by an x-ray diffractometer (BRUKER, D8 Advance) with Cu-Kα radiation (λ = 1.540 56 Å, 40 kV/40 mA) at room temperature. The foam morphologies were studied by using optical microscopy (OLYMPUS-MX63, Japan). Both Cu-matrix and the interface of Al/Cu bimetallic foams were characterized by a JEOL JSM-6510LV electron microscope (Japan) in conjunction with energy-dispersive x-ray spectroscopy (EDS). The mechanical property of cylindrical samples was carried out on a 20KN Servohydraulic universal testing machine (Xinghuo XHTL-Z20KN, Jinan, China) with the loading rate was 0.001 s\(^{-1}\) at room temperature [25]. The ground and polished Cu-matrix surfaces of each specimen were used for measuring the Vickers hardness with a loading weight of 1.96 N (Hengyi VH-5, Shanghai, China). Every sample was measured in 5 random positions with a loading weight of 1.96 N and holding time of 20 s.

2.4. Methods of analysis
The whole x-ray pattern was employed to accurately fit by Rietveld software, MAUD for diffraction information, including background parameters, structural parameters, microstructural parameters, peak shape, specimen parameters, width parameters and quantitative phase analysis, etc [26, 27]. Instrumental parameters obtained from the Si standard sample [28, 29], supposed that the size and strain without broadening, have been introduced in the software [30]. All the variable parameters are refined by introducing an iterative least-squares procedure through the minimization of the residual parameter [27]. The Pseudo-Voigt function is regarded as the primary consideration of diffraction profiles modeling in this method. Cauchy and Gaussian type functions can well model the crystallize size and microstrain broadening, respectively, which are considered as the main causes of the formation of XRD peak profiles broadenings. Both crystallite size and microstrain values were evaluated by adopting the Delf line broadening model of isotropic ‘sizes-train’. The refinement quality of samples in the present work obtained from the analysis all were within the evaluation standard, reaching the weighed residual error (Rwp) < 6.0%, the expected error (Rexp) < 5.0% and the goodness of fit (GoF) < 1.5.

3. Results and discussion
3.1. Foam structure
The outward appearances of the Al-matrix and Cu-matrix in the Al/Cu bimetallic foams (2#) present layer gradients, as shown in figures 2(a-1) and (b-1). The section structure obtained by optical microscopy are shown in figures 2(a-2) and (b-2), in which the pores are attached to each other in both Al-matrix and Cu-matrix. These pores originate from the decomposition of carbamide particles and form a consecutive three-dimensional network during sintering. The formed pore geometry primarily hinges on the distribution of space holder particles, presenting a few millimeters in diameter. Meanwhile, some isolated pores can be found, which formed from the space holder particles discharge through the micropores.
To obtain the planar porosity, the ImageJ software was employed by utilizing both the section image of the Al/Cu bimetallic foams and the binarization method [31]. The appropriate thresholds are chosen and the binarized micrograph picture composed of only black (referring as pores) and white (represents metal matrix) contrasts, as shown in figures 2(a-3) and (b-3). The results evaluated from the binarized micrographs of Al-matrix and Cu-matrix indicate that the proportion of the planar porosities are 50.5% and 55.8%, respectively. The measured porosity of Cu-matrix is higher than that of Al-matrix, and both are larger than the volume of the carbamide particles (48.0 vol%), which can be attributed to the decomposition and dehydration of aluminum dihydrogen phosphate. Therefore, the formed porosities and the structure of pores are dependent not only on the arrangement of carbamide particles but also on the distribution of aluminum dihydrogen phosphate.

3.2. The microstructure of Cu-matrix

The evolutions of the XRD patterns of $\alpha$-Cu (S.G./Fm-3m[225]) [23] as a function of CuSn$_{33}$ content for Cu-matrix are indicated in figure 3. It is shown that the diffraction peaks of $\alpha$-Cu significantly shifts to the lower angle with the increase of CuSn$_{33}$ content, which means more Cu atoms are replaced by larger radius Sn atoms. However, the trace amounts of electron compound $\delta$-phase with a complex cubical lattice (S.G./F-43m[216])

![Figure 2. The Al/Cu bimetallic foams sample 2# for (a) Cu-matrix and (b) Al-matrix with (−1) Morphology, (−2) pore structure micrograph and (−3) binarized micrograph.](image-url)
appears when the content of the CuSn33 increase to 30.0 wt%, which is consistent with the maximum solubility of Sn-Cu is approximately 9.1% [32].

In order to further investigate the effects of CuSn33 content on the microstructure of Cu-matrix, all the Cu-matrix were analyzed with Rietveld method using software of maud [30]. The results of quantitative analysis by Rietveld method for sample 5# is shown in figure 4(a), which shows that the content of α-Cu and δ-phase are 80.96 wt% and 19.04 wt%, respectively. The lattice parameters of α-Cu increase significantly from 3.602 to 3.697 Angstroms with increasing of CuSn33, as shown in figure 4(b). The relation chart between lattice parameters and

![Figure 3. XRD pattern change of Cu-matrix in the Al/Cu bimetallic foams with the increment of CuSn33 content.](image)

![Figure 4. Rietveld refinement results of the Cu-matrix in the Al/Cu bimetallic foams: (a) Phase quantitative analysis of sample 5#, (b) Lattice parameters, (c) Crystallize size and (d) microstrain.](image)
the CuSn33 content is shown in figure 4(b). It shows that the content of CuSn33 is roughly linear to the lattice parameter and the linear fitting equation is $y = 0.00313x + 3.607$.

The crystallite size and microstrain values of $\alpha$-Cu as a function of CuSn33 content are shown in figures 4(c) and (d), respectively. It can be seen that the calculated crystallite size decreases sharply when the 5.0 wt% of CuSn33 was added in Cu-matrix, then decreases slowly and almost keeps the same level size as the content of CuSn33 increases. The crystallites are refined by the addition of CuSn33, which can be explained that the interfacial migration is hindered by the dissolution of Sn atoms into solid solution. However, It is evident that the microstrain of $\alpha$-Cu in Cu-matrix increases gradually until the content of CuSn33 reaches 20 wt% and then drops dramatically as the CuSn33 content continues to increase, as shown in figure 4(d). The microstrain increases with the augment of CuSn33 content, which may come from the solid soluble Sn with larger radius and the concentration of the solid solution increases gradually. Then the solid solution strengthening arise spontaneously due to the elastic interactions between the local stress fields of solute atoms and surrounding dislocations [33, 34]. The increased microstrain produced by all kinds of solute-solvent atoms, which would suppress the dislocation motion more strongly than that in the pure metals or common alloys. From this, the mechanical strength of Cu-matrix can also be improved effectively with regard to the same grain size [33, 34]. However, the microstrain value decreases rapidly when the content of CuSn33 reaches 30.0 wt%. The reason for this is that the precipitated $\delta$-phase formed from the supersaturated solute elements, as a consequence, some of these dislocations rearrange themselves [32].

The pore surfaces morphologies of the Cu-matrix obtained by SEM as function of the CuSn33 content are shown in figure 5. It can be seen from figures 5(a-1) that the Cu powder particles in the sample 1 # almost remains the irregular pressing deformation shape after the heat treatment. Some larger particles appear due to the consolidation of powder particles and the shape of the secondary particles surface is gradually rounded with the increase of CuSn33 content as shown in figures 5(b-1) and (c-1). The results suggest that the additional CuSn33 content with low melting point is conducive to promote the recrystallization of powder particles in the Cu-matrix under the same conditions of heat treatments. The joining mechanism is attributed to the diffuse reaction between Cu and CuSn33, lead to the powder particles inlay each other in the joint. The element distribution of Cu and Sn obtained from the EDS is shown in figure 5. It can be obtained from the EDS analysis that the concentration of Sn elements in particles becomes obvious as shown in figures 5(b-3) and (c-3) with the increasing of CuSn33 content. The thermal diffusion occurs and the substitutional solid solution is formed during the sintering process when the CuSn33 particles were added in Cu-matrix.

### 3.3. Analysis of interface microstructure

The XRD patterns of interfaces for the Al/Cu bimetallic foams (1#, 3# and 5#) as a function of the CuSn33 content are indicated in figure 6. The sample 1# includes three crystalline phases, i.e., $\alpha$-Al (Fm-3m [No.225]), $\alpha$-Cu (Fm-3m [No.225]) and Al$_2$Cu (14/mcm [No.140]) [35]. It also can be found that from the samples 3# and 5# the diffraction peaks of both Al$_2$Cu and Sn (I41/amd [No.141]) [23] become gradually obvious with the content of CuSn33 increasing.

It can be observed from figure 7 that the amounts of Al atoms diffusing to the Cu-matrix are extremely lower compared to the amounts of Cu atoms diffusing to the Al-matrix. The reason is that the diffusion coefficient of Cu in Al is higher than that of Al in Cu [36]. The diffusion occurs in the contact points between the Cu-matrix and Al-matrix interface when the sample is successively subjected to the uniaxial pressure and the sintering process. As the Al-Cu binary phase diagram is known, the contact reaction is triggered to form a liquid at the Al/Cu interface when the temperature reaches 548.2 °C. Once the nucleation of eutectic liquid is completed, both Cu and Al atoms quickly diffuse into the liquid since diffusion is faster in liquid than in solid. Then the liquid contact parts increases and the produced liquid moves along the surface [37, 38]. When the sintering temperature decreases to room temperature, the liquid would transform into the eutectic structure (Sn + Al$_2$Cu). In previous studies of diffusion in friction welded Al/Cu bimetallic joints, some authors reported the formation of Al$_4$Cu$_9$ [39]. It is noted that the Al$_4$Cu$_9$ phase has not been obtained in the present work, which can be explained that Al$_4$Cu$_9$ is generated through the solid-state phase transformation from Al$_2$Cu. However, the diffusion coefficient of the atom in the solid is much smaller than that in the liquid, and the diffusion time is insufficient [21]. Furthermore, the Al-Sn is a simple eutectic system with restricted solid solubilities [40]. The provision of Cu atoms comes from the Cu-matrix may be hindered by the presence of Sn atoms in the liquid phase, i.e., it is difficult to increase the chemical composition of Cu element in the liquid. The remained Sn in the form of segregation principally situates near the interface of Al$_2$Cu and Cu-matrix as shown in figures 7(b-4). During thermal treatment, the reactions of Cu, CuSn33 and Al at the Al/Cu bimetallic foams interfaces can be described by the following equations:

$$\text{Cu(s)} + \text{CuSn}_{33}(s) + \text{Al(s)} \rightarrow (\text{Cu, Al, Sn})$$ (1)
where \( s \) and \( l \) indicate that the solid and liquid phase in the reactants and products, respectively. Then the diffraction peak intensities of both \( \text{Al}_2\text{Cu} \) and Sn simultaneously increase with the increment of \( \text{CuSn}_{33} \) content, which can be explained from equation (2).

### 3.4. Mechanics property analysis

The compressive stress-strain curves of the Al/Cu bimetallic foams without radial constraints are shown in figure 8(a). The diagram shape of the compressive curves can be divided into five typical regions, including initial quasi-linear elastic region (I), followed by a plateau region (II), the second quasi-linear elastic region (III), the second plateau region (IV) and finally, the densification region (V), as indicated in figure 8(b). It can be found that the second collapse plateau stress is lower than the first plateau stress in sample 1, which can be attributed to the unsatisfactory sintering effect in pure Cu-matrix. The average values of the second plateau stress were calculated according to the equation:

\[
\sigma_{\text{plateau}} = \frac{1}{\varepsilon_2 - \varepsilon_1} \int_{\varepsilon_1}^{\varepsilon_2} \sigma \, d\varepsilon
\]

where the 0.28 and 0.50 were chosen as the \( \varepsilon_1 \) and \( \varepsilon_2 \) values in present work, respectively. The average stresses growth rate of the second collapse plateau augments at first with the \( \text{CuSn}_{33} \) content increases as shown in table 1, and then decrease as the content of \( \text{CuSn}_{33} \) continue to grow. The collapse plateau stresses corresponds

\[(\text{Cu}, \text{ Al}, \text{ Sn}) (l) \rightarrow \text{CuAl}_2(s) + \text{Sn}(s)\] (2)
to the plastic buckling of the struts in the foams [4]. The strength of solid solution is increased with the augment of solute content, however, the plasticity and fracture toughness would be relatively reduced, which is the reason that the strength and toughness are generally mutually exclusive [41]. This may be why the average value of the second plateau stress shows a slight drop when the CuSn$_{33}$ content increases to 20.0 wt%. Furthermore, the decreased second collapse plateau stress of CuSn$_{33}$ the content is 30.0 wt%, which can also be attributed to the drastic reduction of microstrain when the δ-phase formed.
The metallic foams with a single component is usually characterized by three stages, i.e. linear elastic deformation, collapse plateau and densification region \[42\]. The present Al/Cu bimetallic foams with two quasi bilinear phases and two plateau regions show an entirely different stress-strain behavior compared with the only one component of metallic foams. It can be obtained that the initial quasi-linear elastic region (I) depends on the elasticities of both Al-matrix and Cu-matrix. The first plateau regions for all compressive stress-strain curves remain at the same level, which can come down to the Al-matrix in samples have the porosity in common. The appearance of the second quasi-linear elastic region (III) can be attributed to two reasons: one is the elastic deformation of Cu-matrix and the second is the densification region of Al-matrix with increasing compressive stress during the same compressive process. As the Cu-matrix is deformed beyond the strain that elastic deformation persists, the stress is no longer proportional to strain, and then the second plateau region (IV) occurs. The densification region of Al-matrix would be hard to sustain since the compressive stresses remain nearly constant during the second collapse plateau phase. The collapse begins to occur at the weakest cross-section in Cu-matrix and spread progressively into the adjacent area until the densification stage (V) of the Al/Cu bimetallic foams happens.

**Figure 8.** (a) The compressive stress-strain curves of the Al/Cu bimetallic foams as function of of CuSn33 content; (b) The characterization of the compressive curves.

| The CuSn33 content | 0.0 wt% | 5.0 wt% | 10.0 wt% | 20.0 wt% | 30.0 wt% |
|--------------------|---------|---------|----------|----------|----------|
| Average plateau stress/Mpa | 12.38   | 22.60   | 28.32    | 27.97    | 23.18    |
| Growth rate of average plateau stress/ % | 82.52   | 128.75  | 125.91   | 87.20    |

The metallic foams with a single component is usually characterized by three stages, i.e. linear elastic deformation, collapse plateau and densification region \[42\]. The present Al/Cu bimetallic foams with two quasi bilinear phases and two plateau regions show an entirely different stress-strain behavior compared with the only one component of metallic foams. It can be obtained that the initial quasi-linear elastic region (I) depends on the elasticities of both Al-matrix and Cu-matrix. The first plateau regions for all compressive stress-strain curves remain at the same level, which can come down to the Al-matrix in samples have the porosity in common. The appearance of the second quasi-linear elastic region (III) can be attributed to two reasons: one is the elastic deformation of Cu-matrix and the second is the densification region of Al-matrix with increasing compressive stress during the same compressive process. As the Cu-matrix is deformed beyond the strain that elastic deformation persists, the stress is no longer proportional to strain, and then the second plateau region (IV) occurs. The densification region of Al-matrix would be hard to sustain since the compressive stresses remain nearly constant during the second collapse plateau phase. The collapse begins to occur at the weakest cross-section in Cu-matrix and spread progressively into the adjacent area until the densification stage (V) of the Al/Cu bimetallic foams happens.
This is confirmed by the images captured at different compressive test stages of sample 4, as shown in figure 9. It can be observed that there is no deformation occurs in both Al-matrix and Cu-matrix in the initial elastic region (I). The distortion occurs initially at the boundary of the Al-matrix side, as can be observed from figure 9(b). It can be explained that the generated Al$_2$Cu with lower plasticity and ductility at the interface of Al/Cu bimetallic foams. That is, a foam specimen starts to deform in the weakest region due to the defects and structural imperfections. Compared with Cu-matrix, the global deformation prematurely occurs in the Al-matrix with increasing in the compressive strain, as indicated from figures 9(b) to (f). The deformation in the Al-matrix leads to the formation of the first plateau region (II), suggesting that the load-bearing capacity of Al-matrix is lower. The first deformation bands are observed at the top of the Cu-matrix, and the deformation bands are oriented in different directions and are inclined at different angles, as shown in figures 9(g) and (h). Then the Cu-matrix foams start to deform in several weakest areas, showing one or more deformation. The progressive collapse of the cells occurs in localized deformation bands, while the cells outside remain in their original shape. These deformation bands continue to deform until local hardening occurs, prompting other deformation bands to initiate collapse [42]. Thereout, the second plateau region (IV) caused by Cu-matrix and the densification region (V) appears successively during the process of compressive. Obviously, the second plateau stresses that arise from Cu-matrix are higher than the first plateau stresses of Al-matrix, which indicated that the plateau stresses present gradient growth and would be benefit to enhance the energy absorption capacity.

The total elastic moduli were determined by measuring the slopes of the initial linear part of the stress-strain curves, as shown in figure 10(a). It can be obtained from table 2 that the values of elastic modulus increase sharply when the CuSn$_{33}$ content is 5.0 wt%, and then keep growing slowly with the argument of the CuSn$_{33}$ content, even decrease as the CuSn$_{33}$ content increases to 30.0 wt%. The elastic moduli decided jointly by Al-matrix and Cu-matrix since these two parts are installed in series as an integral for the Al/Cu bimetallic foams. The compressive stresses loading on both Al and Cu-matrix are equal:

$$\sigma_{Al} = \sigma_{Cu} = \sigma$$  \hspace{1cm} (4)

and

$$l_0 = 2l_{Al0} = 2l_{Cu0}$$  \hspace{1cm} (5)

According to the Hooke’s Law:

$$\sigma(\varepsilon) = E \varepsilon = E \left( \frac{\delta l}{l_0} \right)$$  \hspace{1cm} (6)

The equation can be obtained:

$$\sigma(\varepsilon) = \frac{E_{Al}(\delta l_{Al} + \delta l_{Cu})}{l_{Al0} + l_{Cu0}}$$  \hspace{1cm} (7)

Figure 9. The compression process images captured at five different compressive stages of sample 4 #: (a) Stage I; (b)–(d) Stage II; (e), (f) Stage III; (g), (h) Stage IV; (i) Stage V.
Then equations (4), (5), (7) are solved simultaneously with (8) and (9) can be obtained as follow,

$$E_{\text{Total}} = \frac{E_{\text{Al}}E_{\text{Cu}}}{E_{\text{Al}} + E_{\text{Cu}}}$$  \hspace{1cm} (10)

where $\sigma$, $E$, $l_0$, $\delta l$ and $\varepsilon$ correspond to the compressive stress, elastic modulus, the initial length, the change in length and compressive strain of the Al/Cu bimetallic foams, respectively. From equation (10), it can be obtained that the total elastic modulus of Al/Cu bimetallic foams will be less than the elastic modulus values of either of metals matrix. As one of the elastic modulus increases, the total elastic modulus will increase. Suppose that the elastic modulus values for Al-matrix are equal since all the Al-matrix have the same composition and porosity, the Cu-matrix elastic modulus should increase when the total elastic modulus increases along with the increase

Table 2. The total elastic moduli and the growth rates of the total elastic moduli (relative to the value of CuSn$_{33}$ content is 0.0 wt%) with increasing the CuSn$_{33}$ content.

| CuSn$_{33}$ content | 0.0 wt% | 5.0 wt% | 10.0 wt% | 20.0 wt% | 30.0 wt% |
|---------------------|--------|--------|--------|--------|--------|
| Total elastic modulus/MPa | 245.73 | 588.44 | 625.35 | 731.09 | 670.56 |
| Growth rate of total elastic modulus/% | \( \downarrow \) | 139.47 | 154.49 | 197.52 | 172.89 |

![Figure 10. The total elastic modulus obtained from the slope of the initial linear part of the stress-strain curve with the increment of CuSn$_{33}$ content.](image)
of the CuSn₃₃ content. The increased elastic modulus for Cu-matrix mainly due to the re

fi

ned crystallite and the

surge in microstrain. However, the total elastic modulus shows a decreasing trend when the content of CuSn₃₃ reaches to 30.0 wt%. Such results may be attributed to the dramatical drop of microstrain of Cu-matrix during the generation of δ-phase.

The absorbed energy density of Al/Cu bimetallic foams as a function of the CuSn₃₃ content is shown in figure 11, which are calculated according to the equation:

\[ U(\varepsilon) = \int \sigma(\varepsilon) d\varepsilon \] (11)

The energy absorption density of each sample is defined as the area below the compressive stress curve for up to 50.0% strain [25, 43]. Energy absorption capacities of the Al/Cu bimetallic foams remain pretty much the same with the increment of CuSn₃₃ content when the strain rate approximately are less than 0.25, which rely mainly on the first plateau region stress (II) arise from Al-based. With the increase of CuSn₃₃ content for Cu-matrix, the energy absorption density of 50.0% strain rate increases, especially when the CuSn₃₃ content are 10.0 wt% as shown in table 3. The increased energy absorption density is mainly resulted from the higher second plateau stress, which should be attributed to the fact that the added CuSn₃₃ content improve the microstructure of Cu-matrix. However, the absorption energy density is reduced slightly when the CuSn₃₃ content is 20.0 wt%, which may be attributed to the plasticity and fracture toughness is reduced as the strength of solid solution is increased [41]. The reduction of absorption energy when the CuSn₃₃ content is 30.0 wt% may also be attributed to the produced δ-phase with low toughness and the reduced sharply microstrain of Cu-matrix [22]. As a result, choosing the proper component of CuSn₃₃ in Cu-matrix foams can result in the higher energy absorption capacity with respect to pure Cu-matrix.

The Vickers hardness of the Cu-matrix zone shown in figure 12 was measured to investigate the effect of CuSn₃₃ content on the microhardness of Cu-matrix. The observed hardness value first increases and then remains almost at the same level with the increase of CuSn₃₃ content. Both the average microhardness and the growth rates (see in table 4) become higher with the CuSn₃₃ content increase, which can be attributed to the introduced smaller crystallize and the higher microstrain. The increased hardness value of Cu-matrix becomes slow as the CuSn₃₃ content keeps growing. This may due to the inconspicuous changes of both crystallize size and the microstrain in Cu-matrix. The excessive CuSn₃₃ content would produce high strength δ-phase, but the microstrain drops dramatically at the same process, from this, the microhardness has changed little when the CuSn₃₃ content is 30.0 wt%. It can be obtained from table 4 that the variation trend of microhardness in Cu-matrix is slightly different from that of the elastic modulus, which can be attributed to the microhardness is
related to not only the elastic modulus but also the capacity of local energy dissipation or plasticity [44]. Thus the appropriate content CuSn33 were mixed in the Cu powder particles would form solid solution strengthening due to the occurred diffuse reaction between the Cu and CuSn33 during the sintering process. The mechanical properties of Al/Cu bimetallic foams such as the total elastic modulus, the second plateau stress and the energy absorption density as well as microhardness of Cu-matrix are remarkably improved due to the refined crystallize and the increased microstrain in Cu-matrix.

4. Conclusions

A new type of Al/Cu bimetallic foams with the gradient plateau stresses, consisting of the high strength of Cu and lightweight of Al was obtained by powder metallurgy sintering. The research results suggest the added CuSn33 in Cu-matrix can improve effectively the microstructure and mechanical property of Al/Cu bimetallic foams. The planar porosity results evaluated from the binarized micrographs suggest that the porosity originates from the decomposition of the carbamide particles and aluminum dihydrogen phosphate. The quantitative analysis utilizing Rietveld method shows that the δ-phase forms when the CuSn33 content is 30.0 wt%. The lattice parameter of α-Cu rises linearly as the CuSn33 content increases. The crystallite size of α-Cu first reduces sharply when the CuSn33 was added into Cu-matrix and then decreases slowly with the increase of CuSn33 content. The microstrain of α-Cu increases to the peak value when the content of CuSn33 is the 20.0 wt% and then drops dramatically as CuSn33 content growth continues. The remained Sn in Cu-matrix mainly locates between Al2Cu and Cu-matrix at the interfaces of the Al/Cu bimetallic foams. The compressive stress-strain curves present five typical regions, which is quite distinct from the single component metallic foams with three stages. The second collapse plateau stresses can be elevated by the addition of CuSn33 in the Cu-matrix. Both the elastic modulus and the microhardness were improved obviously when adding CuSn33 in Cu-matrix, which may be due to the refined crystallite and the surge in microstrain of α-Cu. However, the second collapse plateau stresses have not been improved further when the content of CuSn33 reaches to 30.0 wt% due to the generation of low plasticity δ-phase and the falling sharply in microstrain of α-Cu. The produced δ-phase and the fallen dramatically microstrain lead to the microhardness keeps the same level when the CuSn33 content is 30.0 wt%.

Table 4. The average microhardness and the growth rates of average microhardness (relative to the value of CuSn33 content is 0.0 wt%) with increasing the CuSn33 content.

| The CuSn33 content | 0.0 wt% | 5.0 wt% | 10.0 wt% | 20.0 wt% | 30.0 wt% |
|--------------------|---------|---------|----------|----------|----------|
| Average microhardness of Cu-matrix/Hv | 67.20   | 91.00   | 119.54   | 125.16   | 126.56   |
| Growth rate of average microhardness/% | \   | 35.4167 | 77.8869  | 86.25    | 88.3333  |
Acknowledgments

This research work is supported by China’s Sichuan Science and Technology Program (2019YJ0441), Chengdu Normal University First-class Discipline Construction Major Scientific Research Projects (CS18ZDZ03), and the Excellent Student Innovation and Entrepreneurship Training Program of Sichuan Province (S202014389141).

ORCID iDs

Changzhong Liao @ https://orcid.org/0000-0002-7426-870X

References

[1] Cheng Y, Li Y, Chen X, Zhou X and Wang N 2018 Compressive properties and energy absorption of aluminum foams with a wide range of relative densities Mater. Eng. Perform. 27 4016–24
[2] Saruse R, Filippis L A C D, Ludovico A D and Boghetich G 2009 Influence of processing parameters on aluminium foam produced by space holder technique Mater. Des. 30 1878–85
[3] Feng Q, Liao C Z, Ma Y T and Yang G R 2019 Optimization of pore walls microstructure in open cell aluminum foams utilizing self-propagating reaction Mater. Trans. 60 2292–7
[4] Devivier C, Tagliaferri V, Trovaluzzi F and Ucciardello N 2015 Mechanical characterization of open cell aluminum foams reinforced by nickel electro-deposition Mater. Des. 86 272–8
[5] Boomyongmaneerat Y, Schuh C A and Dunand D C 2008 Mechanical properties of reticulated aluminum foams with electrodeposited Ni-W coatings Scr. Mater. 59 336–9
[6] Liu J, Si F, Li D, Liu Y, Cao Z and Wang G Y 2015 Effect of bath pH on electroleless ni-p coating deposited on open-cell aluminum foams Surf. Rev. Lett. 22 1550076 (1–12)
[7] Heness G, Wuhrer R and Yeung W Y 2008 Interfacial strength development of roll bonded aluminum/copper metal laminates Mater. Sci. Eng. A 483–484 740–2
[8] Li X, Zu G and Wang P 2013 Effect of strain rate on tensile performance of Al/Cu/Al laminated composites produced by asymmetrical roll bonding Mater. Sci. Eng. A 575 61–4
[9] Sapanthan T, Khoddam S and Zahir H S 2013 Spiral extrusion of aluminum/copper composite for future manufacturing of hybrid rods: a study of bond strength and interfacial characteristics J. Alloy Compd. 571 85–92
[10] Berski S, Dyja H, Maranda A, Nowaczewski J and Banaszek G 2006 Analysis of quality of bimetallic rod after extrusion process J. Mat. Proc. Techn. 177 582–6
[11] Han J, Wang C Q, Mayer M, Tian Y H, Zhou Y and Wang H H 2008 Growth behavior of Cu/Al intermetallic compounds and cracks in copper ball bonds during isothermal aging Microsc. Rel. 48 416–24
[12] Braunovic M and Alexandrov N 1994 Intermetallic compounds at aluminum-to-copper electrical interfaces: effect of temperature and electric current IEEE Trans. Comp. Pack Manufact. Techn. 17 78–84
[13] Lee W R, Bang K S and Jung S B 2005 Effects of intermetallic compound on the electrical and mechanical properties of friction welded Cu/Al bimetallic joints during annealing J. Alloy Compd. 390 212–9
[14] Rhee K Y, Han W Y, Park H J and Kim S S 2004 Fabrication of aluminum/copper clad composite using hot hydrostatic extrusion process and its material characteristics Mater. Sci. Eng. A 384 70–6
[15] Li X B, Zu G Y, Ding M M, Mu Y L and Wang P 2011 Interfacial microstructure and mechanical properties of Cu/Al clad sheet fabricated by asymmetrical roll bonding and annealing Mater. Sci. Eng. A 529 485–491
[16] Liu W and Cui J Z 1997 The Kirkendall effect of the Al-Cu couple with an electric field J. Mater. Sci. Lett. 16 930–2
[17] Guo Y J, Liu G W, Jin H Y, Shi Z Q and Qiao G J 2011 Intermetallic phase formation in diffusion-bonded Cu/Al laminates J. Mater. Sci. 46 2467–73
[18] Chen C Y, Chen H L and Hwang W S 2006 Influence of interfacial structure development on the fracture mechanism and bond strength of aluminum/copper bimetal plate Mater. Trans. 47 1232–9
[19] Meguro K O M and Kajihara M 2012 Growth behavior of compounds due to solid-state reactive diffusion between Cu and Al J. Mater. Sci. 47 6953–64
[20] Kawakami H, Suzuki J and Nakajima J 2007 Bonding process of Al/Cu dissimilar bonding with liquefaction in air Weld. Int. 21 836–43
[21] Han Y Q, Ben L H, Yao J J, Feng S W and Wu C J 2015 Investigation on the interface of Cu/Al couples during isothermal heating Int J Min Met Mater 22 309–18
[22] Bosco N S and Zok F W 2005 Strength of joints produced by transient liquid phase bonding in the Cu-Sn system Acta Mater. 53 2019–27
[23] Aurelio G, Sommadossi S A and Cuello G J 2012 Crystal structure of Cu-Sn-In alloys around the χ phase field studied by neutron diffraction J. Electron. Mater. 41 3223–31
[24] Aurelio G, Sommadossi S A and Cuello G J 2012 Neutron diffraction study of stability and phase transitions in Cu-Sn-In alloys as alternative Pb-free solders J. Appl. Phys. 112 053520
[25] Standard I, ISO (International Standardization Organization) 13314:2011 (E)(2011) 2011 Mechanical testing of metals—ductility testing—compression test for porous and cellular metals. Ref Number ISO 13314 (13314):1–7
[26] Patra S, Satpati B and Pradhan S K 2009 Microstructure characterization of mechanically synthesized ZnS quantum dots J. Appl. Phys. 106 034313
[27] Lutterotti L and Gialanella S 1998 X-ray diffraction characterization of heavily deformed metallic specimens Acta Mater. 46 101–10
[28] Young R A 1993 The Rietveld method ed R A Young (Oxford UK: IUCr, Oxford University Press) (chapter 1)
[29] Gosh B and Pradhan S K 2009 Microstructure characterization of nanocrystalline Fe3C synthesized by high-energy ball milling J Alloy Compd. 477 127–32
[30] Lutterotti L 2000 Maid: a Rietveld analysis program designed for the internet and experiment integration Acta Cryst. A56 s54
31 Ursula G B and Francis B 2006 A comparison of seven thresholding techniques with the k-means clustering algorithm for measurement of bread-crumb features by digital image analysis J. Food Eng. 74 268–78
32 Viňáš J, Vraeí M, Greš M, Brezina J, Sabadka D, Fedorko G and Molnár V 2018 Restoration of worn movable bridge props with use of bronze claddings Mat. 11 459 (1–13)
33 Wang Z, Fang Q, Li J, Liu B and Liu Y 2018 Effect of lattice distortion on solid solution strengthening of BCC high-entropy alloys J. Mater. Sci. Technol. 34 349–54
34 Duan L, Wu H, Guo L, Xiu W and Yu X 2020 The effect of phase on microstructure and mechanical performance in TiAlN and TiSiN films Mater. Res. Express 7 066401
35 Riani P, Arrighi L, Marazza R, Maione D, Zanicchi G and Ferro R 2004 Ternary rare-earth aluminum systems with copper: a review and a contribution to their assessment J. Phase Equilib. Diffus. 25 22–52
36 Shackelford J F and Alexander W 2001 Materials Science and Engineering Handbook 3rd edn (LLC Boca Raton: CRC Press)
37 Gu X Y, Sun D Q, Liu L and Duan Z Z 2009 Microstructure and mechanical properties of transient liquid phase bonded TiCP/AZ91D joints using copper interlayer J. Alloys Compd. 476 492–9
38 Qian Y Y, Dong Z G and Guo F 2002 Microstructure development and metallurgical analysis of the Al-Si contact reaction J. Mater. Process. Technol. 122 305–8
39 Saeid T, Abdollah-zadeh A and Sazgari B 2010 Weldability and mechanical properties of dissimilar aluminum/copper lap joints made by friction stir welding J. Alloy Compd. 490 652–5
40 ABillur C, Saatci B and Arı M 2020 Thermoelectrical properties of supra-eutectic, eutectic and sub-eutectic compositions of Al-Sn-Mg ternary alloys J. Mol. Struct. 1215 128153
41 Ritchie R O 2011 The conflicts between strength and toughness Nat. Mater. 10 817–22
42 Isabel D, Matej V and Lovre K O 2016 Compressive behaviour of unconstrained and constrained integral-skin closed-cell aluminium foam Compos. Struct. 154 231–8
43 Ibrahim A, Zhang F, Otterstein E and Burkel E 2011 Processing of porous Ti and Ti5Mn foams by spark plasma sintering Mater. Des. 32 146–53
44 Bao Y W, Wang W and Zhou Y C 2004 Investigation of the relationship between elastic modulus and hardness based on depth-sensing indentation measurements Acta Mater. 52 5397–404