Microstructure and shape memory properties of Cu-Al-Fe alloys with different Al contents made by additive manufacturing technology

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Abstract

The Al contents play an exceedingly important role in Cu-Al system shape memory alloys (SMAs), and Cu-Al-Fe alloy represents the new development directions of Cu-Al system SMAs. The Cu-xAl-4Fe (x = 11, 13, 15 wt%) alloys, which take the powder core wire with a structure resistant to element burning as additive manufacturing materials, were prepared by arc melt deposition process. In this work, the as-deposited, quenched and deformed microstructure was studied in detail by utilizing OM, SEM, and XRD. The shape memory properties of the alloys were analyzed by the bending tests. The effect mechanism of the Al content on the shape memory properties of Cu-Al-Fe alloys was also investigated. Results show that the as-deposited microstructure presents sub-eutectic to hyper-eutectic characteristics with the rise in Al content. After quenching, the microstructure of 11 wt% Al, 13 wt% Al, and 15 wt% Al alloys are the \( \alpha' \) martensite, the \( \beta_1' \) martensite, and the \( \beta_1 \) austenite with high order degrees. Under 4% pre-strain, the shape memory recovery rate of the 13 and 15 wt% Al alloy is 100%, but the shape memory recovery rate of 11 wt% Al alloy is only 22.6%. However, compared with the ductility of 11 and 13 wt% Al alloy, that of 15 wt% Al alloy is poor, which causes failure to withstand 4% bending pre-strain. After bending deformation, cracks of 15 wt% Al alloy along the crystals appear and cause the memory strip to break. The analysis indicates that the properties of Cu-Al-Fe alloy have an intense sensitivity to the Al element. The martensitic order degree of the alloy is elevated with the increase in the Al content, and the grain interface gradually becomes sharper. Solidification impurities are formed at the grain boundary during the additive manufacturing process due to the influence of the interface energy. At the same time, the invading \( O_2 \) combines with the more active Al element to form metal oxidation, which markedly reduces the grain boundary strength and the bending strength of the alloy. As a result, the shape memory properties cannot be reflected in the case of high Al content.

1. Introduction

Among the Cu-based shape memory alloys (SMAs), the Cu-Al system alloy has attracted much attention because of its excellent shape memory effect (SME), broad phase transition range and excellent conductivity and thermal conductivity [1, 2]. However, the cold working and mechanical properties of Cu-Al system SMAs are poor due to the polycrystalline brittleness and anisotropy. The SME cannot even be fully exerted owing to its brittleness in work conditions [3, 4]. According to the abovementioned problems, experts have been focusing on the introduction of the third component to developing Cu-Al-X SMAs (X: Ni, Be, Mn, etc). Cu-Al-Ni SMAs have a wide temperature range, which can float between −200 and 200 °C, and are more common in high-temperature applications [5]. Cu-Al-Be and Cu-Al-Mn SMAs were considered to be the closest replacement to Ni-Ti, which is due to their martensitic phase transition, low cost and relatively simple processing. They have also been extensively studied by researchers [6, 7]. Existing research has shown that the Cu-Al-Ni alloys tend to be brittle [8, 9]. The Cu-Al-Be alloys are costly, and the Be element has acute toxicity. The thermostability of Cu-Al-Mn
alloys is also poor [10]. Recently, some scholars have attempted to introduce the Fe element in Cu-Al system SMAs. They found that the Fe element plays a positive role in improving the mechanical properties of Cu-Al system SMAs and the Cu-Al-Fe alloys present excellent SME [11, 12]. For example, Raju et al. [13] adopted the casting method to prepare Cu-Al-Fe SMAs with different Fe contents, and the shape recovery rate of the Cu-12.08Al–2.89Fe alloy is 94% under the pre-strain of 5%.

For Cu-Al system SMAs, the Al content tends to range from 12 wt% to 16 wt% such that a good SME can be obtained. When Al content exceeds 11.5 wt%, the brittleness will increase sharply [14], which limits its application. The addition of the third component Fe will shift the eutectoid transformation point of the alloy to the right. This shift is from the eutectoid transformation point of Cu-Al alloy of 11.8 wt% to the eutectoid transformation point of Cu-Al-Fe alloy of about 13 wt% [14, 15], which expands the application range of Al.

Additive manufacturing SMAs add a fourth dimension, namely, time, based on additive manufacturing ‘3D printing’. This topic can be described as the ‘new darling’ of research in the field of smart materials [16–18]. However, at this stage, the manufacturing method of Cu-Al-Fe SMAs is mostly smelting and casting [19–21], and the additive manufacturing method is mostly used for aluminum-iron bronze [22, 23]. Few types of research have been conducted on the additive manufacturing Cu-Al-Fe SMAs. Thus, investigating them has a great value.

In this work, the Cu-xAl-4Fe (x = 11, 13, 15 wt%) alloys, which take the powder core wire with a structure resistant to element burning as additive manufacturing materials [25], were prepared by arc melt deposition process. The as-deposited, quenched and deformed microstructure was studied by utilizing OM, SEM and XRD. The shape memory properties of the alloys were analyzed by the bending tests. The effect mechanism of the Al content on the shape memory properties of Cu-Al-Fe alloys was also investigated.

2. Materials and methods

2.1. Materials and preparation technology

The powder core wire was prepared by the circular tube method, which contains the preparation of raw materials, grinding of mixed powder, powder filling, and drawing forming, as shown in figure 1. First, the filling powder core was prepared by the design principle of the same composition and same volume’ [26], and the alloy composition was listed in table 1. Secondly, the evenly mixed powder core was filled into the copper tube with the aid of the vibration filling device. Finally, the wire was drawn to a diameter of 2.1 mm using a customized tip-drawing integrated forming device.

The material of the tube was 99.9% T2 copper, and its external and inner diameters were measured as 4 mm and 2.6 mm, respectively. The raw powder used during the preparation was high purity Cu powder (325 mesh, purity of 99.9%), high purity Al powder (200 mesh, purity of 99.9%), and high purity Fe powder (300 mesh, purity of 99.8%).
2.2. Alloy preparation

Figure 2 displays the schematic diagram of the experimental set-up. The additive manufacturing platform adopted the RILAND RC-315SII AC and DC arc welding machine. A sample of approximately 10 mm (in height) × 5 mm (in width) × 80 mm (in length) was deposited on an aluminum bronze substrate (table 2 is chemical composition). The deposition parameters of the experiments included the deposition current of 140 A, the scanning speed of 6 mm s⁻¹, and the argon flow rate of 10 l min⁻¹. Several shape memory strips of 65 mm × 3 mm × 1 mm and microstructure samples with a thickness of 2.5 mm were cut by NSC-M332/W6 wire EDM machine.

2.3. Heat treatment process

The experimental heat treatment included quench and recovery heat treatment after the deformation. The quenching process was 950 °C × 2h → 850 °C × 15 min → water quenching [27], as shown in figure 3. Under normal conditions, the heat treatment process of shape memory recovery after deformation was to heat up to the temperature of Af + 50°C at a heating rate of 15°C min⁻¹ for 10 min.

2.4. Analysis test method of microstructure

The prepared samples were infiltrated in an aqueous solution of FeCl₃ hydrochloric acid for 3~10 s, rinsed with distilled water, and immersed in absolute ethanol for ultrasonic cleaning. The samples were then taken out and dried for microstructure observation. The metallographic microstructure was observed with CMY210 Optical microscopy (OM) and S-3400N scanning electron microscope (SEM). The phase composition of the samples under different conditions was characterized using a D8ADVANCE model of x-ray diffraction (XRD). The XRD testing parameters were Cu target radiation, 40 kV voltage, 75 mA working current, 20°–100° scanning angle, and 0.230 step length. The DSC test was carried out with a synchronous thermal analyzer (STA449F3 Jupiter).

Table 1. Chemical compositions of Cu-Al-Fe SMAs.

| Alloys ID | Cu (wt%) | Al (wt%) | Fe (wt%) |
|-----------|----------|----------|----------|
| 11Al      | Bal.     | 11       | 4        |
| 13Al      | Bal.     | 13       | 4        |
| 15Al      | Bal.     | 15       | 4        |

Table 2. Chemical composition of aluminum bronze plate.

| Element | Mass fraction(%) | Element | Mass fraction(%) |
|---------|------------------|---------|------------------|
| Cu      | Bal.             | Zn      | 0.85             |
| Al      | 9.52             | Sn      | 0.08             |
| Fe      | 3.57             | Si      | 0.07             |
| Mn      | 0.39             | Pb      | 0.006            |
| P       | 0.008            |         |                  |
produced by NETZSCH, Germany. The DSC test parameters were 15°C min\(^{-1}\), a single cycle of cooling and heating, and the cycle temperature was room temperature (RT) – 650°C - RT.

2.5. Test method of SME

The elastic recovery (ER) ratio \(\theta\) (equation (2)) and shape memory recovery (MR) ratio \(\eta\) (equation (3)) were measured by bending tests, as exhibited in figure 4. The mold size \(D\) was 24 mm, and thus the pre-strain \(\varepsilon\) (equation (1)) was 4%. For each test, three samples were employed to obtain measured values and standard error.

\[
\varepsilon = \frac{t}{D + t} \times 100\% \quad (1)
\]

\[
\theta = \frac{\theta_{ER}}{180} \times 100\% \quad (2)
\]

\[
\eta = \frac{\theta_{MR} - \theta_{ER}}{180 - \theta_{ER}} \times 100\% \quad (3)
\]

3. Results and analysis

3.1. As-deposited microstructures

According to the Cu-Al-Fe ternary phase diagram at 4% Fe content [28] (figure 6) and the study of Qinglong Yuan [29], the as-deposited microstructure of the alloy is composed of \(\alpha + \gamma_2 + \kappa\) phases. As shown in figure 5, in the 11Al alloy, the alloy microstructure presents a sub-eutectic structure. Many white \(\alpha\) phases and black layers of the Fe-rich \(\kappa\) phase can be found along the grain synchronous on the substrate of \(\alpha + \gamma_2 + \kappa\) phases, and the white \(\alpha\) phase and the black Fe-rich \(\kappa\) phase alternate can also be observed. In the 13Al alloy, the microstructure is a near-eutectic structure, and the solid solubility of Fe in the eutectic structure increases. Compared with the black Fe-rich \(\kappa\) phases precipitated on the grain boundary of the 11Al alloy, those of the 13Al alloy are reduced and show on the eutectic microstructure substrate instead. As the precipitation phase of Fe at the grain boundary decreases, the white \(\alpha\) phase precipitates on the grain boundary begin to be narrow, thin and discontinuous. In
the 15Al alloy, the microstructure is a hyper-eutectic microstructure, where a large number of the flower \( \gamma_2 \) phases distribute on the grey eutectic microstructure substrate, and the Fe-rich \( \kappa \) phase disappears. At this moment, the existence of grain boundaries is difficult to observe.

As shown in figure 7, the peak strength of 43° decreases with the increase in the Al content, which indicates the reduced proportion of the \( \alpha \) phase and \( \kappa \) phase. \( \alpha \) phase and \( \kappa \) phase tend to disappear even in the 15Al alloy. The peak strength of 44° increases continuously with the rise in the \( \beta_1 \) phase precipitates. According to Bragg’s law, the 43° peak shifts to the left due to the increase in the solid solubility of the Al. The analysis demonstrates that the resulting intermetallic compound may have occurred from the conversion of Fe6Al, Fe3Al to FeAl.

### 3.2. Quenched microstructure

As shown in figure 8, the martensitic structure in the quenched microstructure of 11Al alloy is the lamellar structure. Many white polygonal microstructures can be found, and many \( \kappa \) phases are precipitated in the grain. From the arrangement point of view, the martensitic order degree at this time is poor. In the 13Al alloy, the development of martensitic is uniform in the grain, in which the width of martensitic is large. The arrangement between the martensitic is precise, and the gaps are small. The black dot precipitation in the grain is reduced, and it is accompanied by a small amount of white circular precipitates. In the 15Al alloy, the presence of martensitic structure can no longer be observed, and the grains are full of black granular precipitates. From the point of view of grain size, the grain size of 11Al alloy is thinner with poor uniformity, whilst the grain size of 13Al alloy is the largest amongst the three alloys. The grain size of the 15Al alloy is medium with a high degree of homogenization as well.

Figure 9 shows that the main phase after quenching of the 11Al alloy is \( \alpha' \) martensitic, and the second phase is \( \alpha \) phase and \( \kappa \) phase. For the 13Al alloy, its diffraction pattern shows that the martensitic peak transits from \( \alpha' \) to another martensitic \( \beta_1' \). In addition, the \( \alpha \) and \( \kappa \) phases are present in the alloy, and the proportion of the \( \kappa \) phase rises. In 15Al alloy, the phase diagram shows no presence of martensitic structure after quenching, and a large number of \( \beta_1 \) phases are taken instead. Compared with quenched microstructures, this difference in phases (e.g. phase shifts and content differences) causes the evolution of microstructure. At the same time, the change in Al content also causes the shift of the peak, which is accompanied by the change in the Al/Fe ratio and the shift of the 43° and 45° peaks to the left. Furthermore, the solid solution of the solute atom will increase the crystal plane distance. According to the Bragg equation, the crystal plane distance is inversely proportional to the peak angle. Therefore, the Al/Fe ratio and the solid solution of Fe and Al increase.

The analysis indicates that the abovementioned results are due to the difference in phase transformation during solidification caused by the change in Al content. From the phase diagram of the Cu-Al-Fe alloy with the Fe mass fraction of 4% (figure 6), the composition phase of the 11Al alloy at high temperature is \( \alpha \) and \( \beta \) phases. The \( \alpha \) phase does not change during quenching, but the \( \beta \) phase is transformed into an \( \alpha' \) phase. Therefore, the fine lamellar (\( \alpha' \) phase) structure and white polygonal structure (\( \alpha \) phase) exist in the 11Al metallographic diagram. However, the high-temperature composition phase in the 13Al alloy is a single \( \beta \) phase. During quenching, the \( \beta \) phase is transformed into the \( \beta_1 \) phase (the grey plate microstructure shown in figure 7) by martensitic transformation. In the 15Al alloy, the composition phase of the Cu-Al-Fe alloy at high temperature is also the single \( \beta \) phase. However, the quenching environment in this study has not met the needs of the martensitic transformation region of the alloy. The \( \beta \) phase is transformed into the \( \beta_1 \) phase by an order-disorder transformation. Given the high Al content, Fe fails to inhibit its diffusion. Therefore, \( \gamma_2 \) phases with a large area are precipitated within the grain boundary. The change in Al content changes not only its solid solubility but also the solid solubility of other elements (e.g. Fe), which triggers the changes in the alloy phase in the process of alloy quenching. The results also reflect that the martensitic types obtained by the Cu-Al-Fe alloy
with different Al contents during the quenching are different. The high sensitivity of the martensitic transformation point to the Al content is shown from the side.

### 3.3. Shape memory properties of Cu-Al-Fe alloys

The 11Al alloy is extremely elastic with an ER ratio of 53.5%, while its MR ratio is only 22.6%. As shown in figure 10, the alloy MR ratio boosts from 22.6% to 100% under the pre-strain of 4% when the Al content increases slightly from 11 wt% to 15 wt%. This enhancement also reflects that the shape memory properties of the alloy show high sensitivity and dependence on the Al content in the Cu-Al system SMAs. The 15Al alloy breaks when bending. Thus, its ER ratio is not referenced, but its MR ratio is still as high as 100%. The results indicate that a better balance between shape memory properties and ductility can be achieved in the 13Al alloy, and the alloy remains intact and fully recovers under bending deformation.

### 3.4. Deformed microstructure

Figure 11 shows the deformed microstructure of Cu-Al-Fe alloys with different Al mass fractions. In the 11 Al alloy, the martensite form after the deformation is more orderly. Nevertheless, a crossover phenomenon occurs inside the grain, and the grain orientations in different grains are quite different. In local areas, martensite...
development is interrupted, and the growth is more chaotic at the crystal boundaries. After the deformation of the 13Al alloy, the grain morphology changes from the multi-ribbed band before the deformation to the equiaxed shape, and the martensitic orientation in each grain tends to be consistent. However, the martensite orientation in different grains has different degrees of difference. During the deformation process, the 15Al alloy induces the formation of martensite due to the action of stress. Therefore, not only the intragranular martensite orientation is consistent but also the martensite orientation difference between grains is small (nearly tending to consistent), which results in the transgranular phenomenon of martensite in the local area. At this time, the martensite is in the shape of wide and thick lath, and the grain shape is a straight-sided polygon.

Figure 8. Quenched microstructures of Cu-xAl-4Fe alloys (a) 11 Al, (b) 13 Al, (c) 15Al.

The analysis indicates that the κ phase gradually increases with the decrease in Al content when the Fe content is stable. This phenomenon will hinder the development of martensitic to a certain extent. It will even change the martensite orientation or interrupt its growth in severe cases, which results in disordered orientation. The martensitic of 11 Al alloys are particularly chaotic at the crystal boundary because the κ phase precipitation and quenching defects on the crystal boundaries are the most serious. Martensitic transformation from β phase to β′ phase occur in 13Al alloy during quenching. After the deformation of the grain morphology, the martensitic orientation within each grain is consistent from the deformation of the multi-ribbed band to the equiaxed shape. However, different martensitic order degrees are observed between grains because of the bonding of grain boundaries and the difference in intergrain orientation. Therefore, 15Al alloy goes beyond the martensitic transformation zone during quenching. At this time, the β phase no longer undergoes martensitic transformation and changes to the β1 phase. During cooling, part of β1 changes to γ2. No thermoelastic martensite transformation is observed in the quenching stage. As a result, the martensite inside and outside the crystal can grow orderly along the stress direction. When the Al content is too high, the grain boundary precipitation turns to the intercrystalline. In this case, the grain boundary moves more violently under the same driving force, which is prone to martensite transgranular growth. At the same time, the movement of the grain

Figure 9. X-ray diffraction patterns from quenched specimens.
boundary consumes dislocations. The martensitic order degree at the crystal boundary is improved, and the grain interface gradually becomes sharper.

Figure 12 shows a fracture morphology of the 15Al alloy measuring shape recovery rate, in which figure 12(a) is a low-magnification topography, and figure 12(b) is a high magnification morphology of the outer fracture zone. The bulk grains are visible in the fractures, and the grains are stretched along the memory strip axis. The crystal surfaces rise and fall, and local cracks along the crystals are still visible in figure 12(a). By comparing the two sides of the bend, the fracture along the crystal occurs firstly on the outer side. In the meantime, the inner side is the crack extension zone, and its surface is relatively flat. Figure 12(b) shows that the alloy contains many metal oxides and precipitates.

4. Discuss

4.1. Effect of Al contents on phase transition feature points

According to the alloy deformation and recovery process and the DSC curve of Cu-Al-Fe alloy with different Al contents (figure 13), the phase transition characteristics of Cu-Al-Fe alloy are sensitive to the fluctuations of Al element composition. The change in Al content changes not only the type of martensite in the alloy but also the martensite transformation point. The results in figure 13 show that the martensitic transformation point of the alloy moves to the left (the martensitic transformation start temperature (Ms) decreases) with the increase in Al content. When the Al content increases from 11 wt% to 15 wt% during the process, Ms decreases by up to 217 °C. The downward trend gradually increases during the change, but it does not show a linear relationship with Al content. Ms decreases by 68 °C when 11Al increases to 13Al, whilst Ms decreases by 149 °C when it increases to 15Al.

The analysis demonstrates that the sensitivity continues to increase when the Al content changes, and this phenomenon occurs even if the third component remains unchanged. This finding is due to the stronger sense of the phase transition, which contributes to the increase in Ms phase transition distance. At the same time, the phase transition peak gradually increases with the lifting of the Al content, which implies that the order of the
three alloys gradually increases, and the recovery ability becomes stronger. The explosive recovery that occurs during the recovery process may be related to the martensitic structure, and the high peak strength causes the strong heat absorption capacity and fast speed of shape recovery. All three alloys possess a unified structure during the phase transition, which means that they have only one martensitic involvement when they undergo a thermoelastic phase transition.

4.2. Effect of Al contents on quenching phase transition behavior

In the Cu-Al system SMAs, the Al content has a dominant effect on shape memory properties. When the Al content is between 9.4 and 15.6 wt%, the eutectoid transformation $\beta \rightarrow \alpha + \kappa + \gamma_2$ occurs firstly at 565 $^\circ$C during the cooling process. This phenomenon results in eutectoid ($\alpha + \gamma_2$) or eutectoid and pre-precipitation phase ($\alpha$ or $\gamma_2$) mixed microstructure. When the Al content is between 9.6 and 15.6 wt%, the non-diffusion martensitic transformation occurs during the rapid cooling process from the high-temperature $\beta$ phase. According to the Cu-Al binary phase diagram in figure 14 [30], the Cu-Al-Fe SMA is a mixed $\alpha + \beta$ phase at high temperature when the Al content is too low ($W(Al) = 11\%$ in this study), and the martensite after rapid cooling is a disordered $\alpha'$ phase (or $\beta'$ phase). When the Al content is moderate ($W(Al) = 13\%$), the high-temperature phase of Cu-Al-Fe SMAs is a single $\beta$ phase. The martensite formed by rapid cooling is an 18R structure, which is a $\beta_1'$ phase. It inherits the order of the parent phase. When the Al content is too high ($W(Al) = 15\%$), the high-temperature phase of the Cu-Al-Fe SMAs is a single $\beta$ phase. After rapid cooling, the martensitic phase transition characteristic point is greatly dropped due to the increase in Al content. The martensite transformation temperature is not reached under this process in this study. As a result, no martensite is formed. In summary, high Al content causes the higher order of the alloy and the sharper grain interface.
4.3. Effect of Al contents on grain boundary diffusion

On the one hand, the existence of a large number of defects such as dislocations at the grain boundary and quenching pore positions leads to the diffusion of Al and other elements during the quenching process of Cu-Al alloy. This phenomenon leads to the partial aggregation of precipitates. The bending recovery process shows that, when low Al content and high Fe content appear simultaneously, precipitation exists but with relatively uniform distribution. The precipitate segregation at the grain boundaries is severe with the increase in Al content, and many brittle $\gamma_2$ phases are generated. This phenomenon is also one of the main reasons for the bending fracture of the alloy. On the other hand, the presence of defects and metal oxides enables more active Al elements to combine with oxygen. This phenomenon destroys the strength of the grain boundary and rapidly reduces the bending strength of the alloy. Thus, it fails to reflect its shape memory properties.

Sampling and analyzing the element distribution at the grain boundary of the deformed samples of 11Al and 15Al alloys show that the element distribution at the grain boundary is consistent with that in the crystal part (figure 15(a)) when Al is low (e.g. the 11Al alloy). The Al and O contents at the grain boundary are significantly higher than that in the intragranular part (figure 15(b)) when Al is high (e.g. the 15Al alloy). Therefore, the Al element in Cu-Al-Fe alloy diffuses seriously to the grain boundary when Al is high. At the same time, the Al element ‘absorbs oxygen’ in the additive manufacturing process and segregates at the grain boundary. The combination of the two forms a metal oxide to destroy the grain boundary strength. This condition leads to the increase in brittleness of the alloy and the occurrence of intergranular brittle fracture during deformation, which reduces its practicality.
5. Conclusions

The as-deposited, quenched and deformed microstructure was studied in detail. The martensitic transition and the shape memory properties of Cu-xAl-4 Fe (x = 11, 13, 15 wt%) were analyzed. The conclusions were summarized as follows:

1. In the deposited microstructure, the microstructure of 11Al alloy is mainly composed of the α phase. The microstructure of 13Al alloy is mainly eutectoid, and the microstructure of 15Al alloy is mainly composed of a plum-like γ2 phase. The change law from sub-eutectic to over-eutectic structure is presented with the increase in Al content. In the quenched-state alloy, the change in the mass fraction of Al not only alters its solid solution but also leads to the fluctuations in the solid solution of other elements (e.g. Fe). These phenomena trigger the changes in the process of quenching phase-changing of the alloy. The 11Al, 13Al and 15Al alloys are composed of α′ martensitic, β1′ martensitic and β1 austenite with a higher degree of order. Therefore, the martensitic structure obtained by Cu-Al-Fe alloys with different Al contents in the quenching process differs. The high sensitivity of martensitic phase transition points to Al content is shown from the side.

2. The Al content greatly influences the deformation behavior of Cu-Al-Fe alloy, and ER and MR indicate a strong sensitivity. Under 4% pre-strain, the ER and MR ratios of 13Al alloy are 39.11% and 100%, respectively. The 11Al ER ratio of the alloy is the highest, which reaches 53.48%. However, its MR ratio decreases significantly to only 22.6%. Although the MR ratio of the 15Al alloy is as high as 100%, the ER ratio cannot be measured given that the memory strip is directly broken during the bending process.

3. High Al content causes the great brittleness of the alloy. 11Al and 13Al alloys are highly plastic and remain intact at 4% pre-strain. However, the 15Al alloy cannot withstand 4% bending pre-strain, and cracking along the crystal appears in the curved tissue. The bending fracture of this tissue shows a significant along-crystal fracture morphology. On the one hand, a large number of defects such as dislocations and quenching pores occur at the grain boundary during the quenching process of Cu-Al alloy. This occurrence results in the diffusion of Al and other elements, and then, the precipitates aggregate locally. With the increase in Al element content, the precipitation at the grain boundary is seriously analyzed, and a large number of γ2 brittle phases are generated. On the other hand, the grain boundary is susceptible to O2 invasion in the process of additive manufacturing due to the presence of defects. The combination of the O element and the active element Al forms a metal oxide to damage the grain boundary strength and decrease the bending strength of the alloy rapidly. As a result, its memory performance cannot be fully reflected.

4. The phase transition sensitivity continuously increases when the Al content changes even if the third component remains unchanged. This phenomenon is due to the strong dependence on Al elements. Therefore, the temperature difference between Ms and Ms is enlarged. At the same time, the order of the alloy gradually increases and the recovery ability becomes stronger as the phase transition peak of the Al content gradually increases.

5. This study has conducted detailed research on the microstructure and properties of additive manufacturing Cu-Al-Fe alloys. However, it has not comprehensively explored the forming and heat treatment processes. The forming and heat treatment processes can be studied in the future.

Data availability statement

All data that support the findings of this study are included within the article (and any supplementary files).

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