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Effect of directional solidification rate on the microstructure and properties of deformation-processed Cu-7Cr-0.1Ag in situ composites

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Abstract

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Keywords

properties, rate, solidification, deformation, directional, effect, microstructure, processed, composites, cu, 7cr, 1ag, situ

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Effect of directional solidification rate on the microstructure and properties of deformation-processed Cu-7Cr-0.1Ag in situ composites

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Abstract: The influence of directional solidification rate on the microstructure, mechanical properties and conductivity of deformation-processed Cu-7Cr-0.1Ag in situ composites produced by thermo-mechanical processing was systematically investigated. The microstructure was analyzed by optical microscopy and scanning electronic microscopy. The mechanical properties and conductivity were evaluated by tensile-testing machine and micro-ohmmeter, respectively. The results indicate that the size, shape and distribution of second-phase Cr grains are significantly different in the Cu-7Cr-0.1Ag alloys with different growth rates. At a growth rate of 200 μm·s^{-1}, the Cr grains transform into fine Cr fiber-like grains parallel to the pulling direction from the Cr dendrites. The tensile strength of the Cu-7Cr-0.1Ag in situ composites from the directional solidification (DS) alloys is significantly higher than that from the as-cast alloy, while the conductivity of the in situ composites from the DS alloys is slightly lower than that from the as-cast alloy. The following combinations of tensile strength, elongation to fracture and conductivity of the Cu-7Cr-0.1Ag in situ composites from the DS alloy with a growth rate of 200 μm·s^{-1} and a cumulative cold deformation strain of 8 after isochronic aging treatment for 1h can be obtained respectively as: (i) 1067 MPa, 2.9% and 74.9% IACS; or (ii) 1018 MPa, 3.0%, and 76.0% IACS or (iii) 906 MPa, 3.3% and 77.6% IACS.

Key words: directional solidification rate; Cu-Cr-Ag; in situ composite; microstructure; mechanical properties; conductivity

1. Introduction

Over the past decades, deformation-processed binary Cu-Cr in situ composite has attracted considerable attention due to the relatively economical cost and excellent properties [1, 2]. The recent progress in the electric, microelectronic, energy and automobile industries has increased property requirements, so that conductive materials need to possess good conductivity and high

strength, and be available at reasonable cost. As a consequence, extensive research is now being carried out on various Cu-Cr system in situ composites [3-7]. Previous research [8, 9] has explored two main approaches to improve the strength and conductivity of these in situ composites. One is to modify their processing using an intermediate heat treatment. The other is to modify the composition, particularly using extra alloying elements. Ag has been used as a third element in many studies [3, 6, 7, 9, 10], because the conductivity of Ag is higher than that of Cu, and the electronic structure, crystal structure and electronegativity of Ag are similar to that of Cu.

Recently, in order to broaden the application field of these in situ composites, many other properties such as thermal stability, inoxidizability, ductility and fatigue properties have been studied. Raabe and Ge [10] investigated the thermal stability of Cr filaments in Cu-10Cr-3Ag in situ composite. The result indicated that the filaments underwent a capillarity-driven shape change from bamboo morphology with grain boundary grooves to complete spherodization during annealing. The oxidation behavior of Cu and Cu-10% Cr in situ composite was studied by Haugsrud and Lee [11] at 400-700 ℃ in air and in argon containing 10 ppm O_2. They found that the presence of the Cr fibers increases the oxidation resistance compared to unalloyed copper up to 600 ℃. The work of Zhang et al. [12] showed that the Cu-15Cr-0.2Ti in situ composite failed by first fracturing the chromium fibers to nucleate the microvoids and then rupturing the Cu matrix. The result obtained by Masuda and Tanaka [13] suggested that it was very important to reduce the size and amount of un-dissolved Cr particles in order to improve the fatigue strength of Cu-Cr composite. However, to the present authors' knowledge, even though numerous results have been published on these composites, there is still a lack of systematical investigation reports concerning the possibility of improving the properties of Cu-Cr in situ composites by controlling the microstructure of the as-cast starting ingots.

The as-cast starting microstructure is a key factor influencing the final properties of Cu-Cr in situ composites [14]. Directional solidification (DS) is a promising technique to obtain a desirable microstructure by controlling the growth direction of crystals. The thermal gradient ahead of the solid/liquid interface and the growth rate are two key parameters in the DS process. Since the conventional DS process is difficult to meet the increasing demand for G by modern industry, a

series of DS processes, such as Bridgeman process, gas cooling casting (GCC) and liquid-metal cooling (LMC), have been developed by several researchers in order to obtain a higher G [15,16]. Subsequently, many DS variations are established and applied to prepare various functional and structural materials [17-19].

The developments of DS processes have made significant contributions to achieving ultra-fine columnar crystals, and to investigating solidification theories. However, there are still few researches focusing on the deformation-processed in situ composites under DS. For this reason, a study was carried out in the current work to examine the microstructural characteristics and properties as functions of growth rates of the deformation-processed Cu-7Cr-0.1Ag in situ composites prepared by using DS process.

2. Experimental details

The Cu-7Cr and Cu-7Cr-0.1Ag alloys were prepared by melting the appropriate amounts of electrolytic Cu, commercial Cr and Ag (of at least 99.94 wt.% purity) in a magnesia crucible using a vacuum induction furnace. The molten alloys were casted into rod shaped ingots of about 36 mm in diameter by using a graphite mold. Bars of 6.8 mm in diameter and 100 mm in length were machined from the as-cast ingots. Each bar was placed into an alumina crucible, which had been coated with yttria to isolate the alloy from the crucible. A Bridgeman type apparatus was employed to prepare the directionally solidified bars under the protection of high purity argon at 380 Pa. After the alloys were melted, homogenised by holding for 20 min, the bars were directionally solidified at growth rates of 5 - 200 $\mu m \cdot s^{-1}$.

The in situ composites were produced by thermo-mechanical processing as follows. The bars were heated up to 950 ℃ at a rate of 5 K/min, held for 70 min and then water quenched. The quenched bars were cold drawn to a cumulative cold deformation strain of $\eta = 7$. Then they were subjected to an intermediate heat treatment by heating to 500 ℃ at a rate of 5 K/min, held for 1 h and then furnace-cooled to room temperature. These heat-treated bars were cold drawn to $\eta = 8$, and aged by heating to temperatures from 200 - 600 ℃ at a rate of 5 K/min, held for 1 h and finally furnace-cooled to room temperature. The cumulative cold deformation strain was obtained by:

$$\eta = \ln(A_0 / A_f) \quad (1)$$

where A_0 is the original cross-sectional area, and A_f is the final cross-sectional area.

A Leica DMI5000 M optical microscope (OM) and a JSM-6360LV scanning electronic microscope (SEM) were employed to characterize the microstructure of the as-cast and deformation-processed specimens. The OM and SEM specimens were prepared through mounting, mechanical grinding, polishing and then etching in a solution of 120 ml H_2O, 20 ml HCl and 5 g $FeCl_3$. An EMT2203-B electronic tensile-testing machine was used to evaluate the tensile properties of the deformation-processed specimens. The ultimate tensile stress (UTS) was taken as a measure of the tensile strength for comparison purposes because it was very reproducible and well defined for similar specimens. Each tensile strength value was calculated from an average of at least eight measurements with deviation being within 4%. A ZY9987 digital micro-ohmmeter with precision of 1 μΩ was employed to measure the electrical resistivity (ρ) at room temperature. The corresponding conductivity was evaluated according to the definition of International Annealed Copper Standard (ICAS) in which 1.7241μΩ·cm is defined as 100% IACS.

3. Results and discussion

3.1. Microstructure

Fig. 1(a) presents the microstructure of the as-cast Cu-7Cr-0.1Ag alloy. It can be seen that the second phase Cr dendrites (black) are embedded in the copper matrix and randomly oriented with the bar axis of the as-cast alloy. Fig. 1(b), (c) and (d) show the microstructure of the DS Cu-7Cr-0.1Ag alloys at different growth rates. As observed in Fig. 1, the size, shape and distribution of the second-phase Cr grains are significantly different in the material processed from DS bars. At a growth rate of 5 $\mu m \cdot s^{-1}$, the Cr grains of the DS Cu-7Cr-0.1Ag alloy still show dendritic character, and the grain size is smaller as compared to that of the as-cast alloy, as shown in Fig. 1(b). The Cr dendrites proportion declines gradually with the increase of the growth rate. At a growth rate of 50 $\mu m \cdot s^{-1}$, the Cr dendrites disappear and form Cr fiber-like grains parallel to the pulling direction. However, as shown in Fig. 1(c), the gains are relatively coarse and distribute non-uniformly in the matrix. Further increasing the growth rate to 200 $\mu m \cdot s^{-1}$, the Cr fiber-like

grains are finer relative to that at other growth rates and distribute uniformly, as shown in Fig. 1(d). Similar results are obtained when DS is employed to produce the Cu-7Cr alloys. These are attributed to the DS technique, which produces a strict one-dimensional thermal flux along the sample axis and sufficient heat exchange between the sample and cooling medium by an appropriate temperature gradient and growth rate [16].

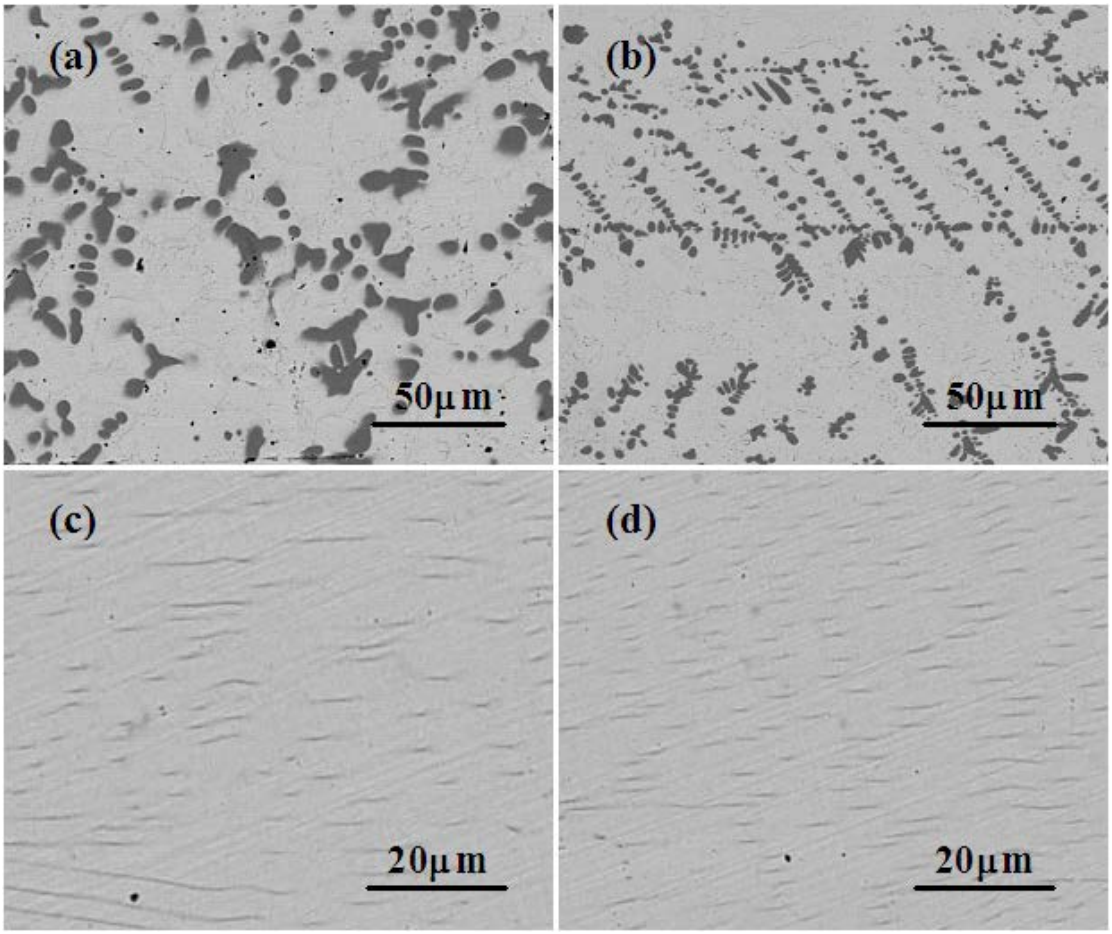

Fig. 1 Longitudinal sectional SEM microstructures of the Cu-7Cr-0.1Ag alloys: (a) as-cast; (b) DS at a growth rate of 5 $\mu m \cdot s^{-1}$; (c) DS at a growth rate of 50 $\mu m \cdot s^{-1}$; (d) DS at a growth rate of 200 $\mu m \cdot s^{-1}$

According to the binary equilibrium phase diagram of Cu and Cr [20], there is a eutectic point at 1.28% Cr. Subsequently, the concentration gradient of the liquidus of hypereutectic primary β phase increases and the crystallization temperature interval widens with increasing the Cr content, resulting in a relatively broad pasty solidification zone. At a relatively low growth rate, the solidification rate of those Cu-Cr alloys is slow, and also the nucleation rate of the primary Cr

phase is slow. The nucleated Cr phase has relatively long time to grow in the broad pasty solidification zone, which is beneficial to Cr phase to develop dendrites. With the increase of the growth rate, the non-equilibrium α phase starts to precipitate from the communal zone of the primary Cr and liquid phases, and depends on the primary Cr dendrites to nucleate and grow which will impede the growth of the Cr dendrites. Further increasing the growth rate, the nucleation rate of the primary Cr phase increases, and the relatively high cooling velocity may cause insufficient time for diffusion and growing, which will impede the free growth of Cr phase and promote it to grow along a strict one-dimensional thermal flux to form DS Cr fiber-like grains. According to the solidification theory, the heat-sinking condition of the solidification interface is the key factor of the solidification microstructure. A higher heat-sinking capability results in a smaller range of influence produced by the crystallization of the primary Cr phase, which is beneficial to the refinement of the primary Cr phase [21]. The dendrite growth theory of Trivedi [22, 23] indicates, when the growth undercooling is small in DS process, the growth rate and curvature radius of dendrite tip meet the following relationship:

$$r = 2GD^2(1-k_0)/[k_0R(\Delta T_0 R - GD)] \quad (2)$$

where r is the curvature radius of dendrite tip, G is the thermal gradient of dendrite tip, D is the diffusion coefficient of liquid solute, k_0 is the distribution coefficient of solute, R is the growth rate, and ΔT_0 is the crystallization temperature range. The curvature radius of dendrite tip decreases with increasing the growth rate, as indicated in Eq. (2). Under certain solidification conditions, the grain size, shape and distribution are determined by the values of R, G, solid/liquid interfacial energy and solute distribution of the growth interface. With the increase of R, the solute transport dynamic balance of the solid/liquid interface front will be destroyed, leading to the instability of the original shape of solid/liquid interface. In order to achieve a new dynamic balance of solute transport and growth interface morphology, the only way is to reduce the curvature radius of the growth front. Therefore, the increase of growth rate results in finer Cr phase at the constant other solidification parameters.

Fig. 2 presents typical microstructures of longitudinal sections of the deformation-processed in situ composites with a cumulative cold deformation strain of η = 4 from the as-cast Cu-7Cr and

Cu-7Cr-0.1Ag alloys, and the DS Cu-7Cr and Cu-7Cr-0.1Ag alloys with a growth rate of 200 $\mu m \cdot s^{-1}$. The four in situ composites exhibit similar microstructural characterization, i.e., all consist of a Cu matrix, elongated Cr grains and/or thin Cr fibers parallel to the drawing direction. In the transverse sections of these in situ composites, the second phase typically is rendered into a curled state [9, 24, 25]. This is attributed to the deformation producing Cr fibres with a <111> fibre texture, which promotes plane strain deformation rather than axially symmetric flow. However, the Cu matrix does deform in an axially symmetric manner during wire-drawing, which constrains and forces the Cr fibres to fold or twist about the wire axis to maintain compatibility with the matrix and produces the irregular cross-section [9]. Differently, the deformation of the Cr dendrites of the as-cast Cu-7Cr distributes non-uniformly, and there are still many round Cr grains, as shown in Fig. 2(a). In contrast, the Cr fiber-like grains of the DS Cu-7Cr with a growth rate of 200 $\mu m \cdot s^{-1}$ are all drawn into long thin fibers, as compared in Fig. 2(a) and (b). Similar results are also obtained in the Cu-7Cr-0.1Ag, as shown in Fig. 2(c) and (d). This is attributed to the DS technique that produces finer and more uniform Cr phase and forms Cr fiber-like grains elongated parallel to the drawing direction. The microstructure of the deformation-processed Cu-7Cr-0.1Ag in situ composites is finer than that of the Cu-7Cr processed with the same cumulative cold deformation strain, respectively. This is attributed to the initial finer dendrite size in the as-cast or DS Cu-7Cr-0.1Ag alloy and the better co-deformation of multiphase during the deformation-driven microstructure evolution due to a stronger Cu matrix, strengthened by Ag in solid solution [9, 26].

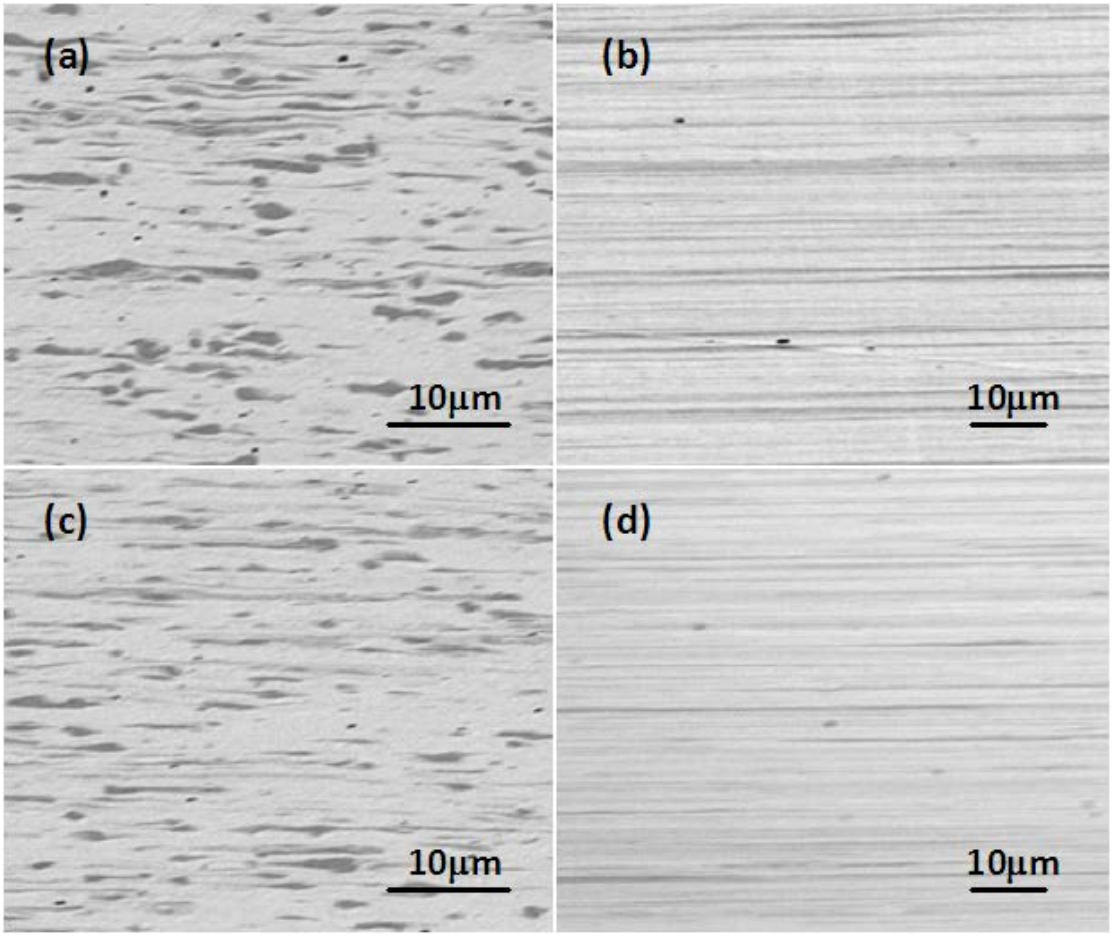

Fig. 2 Longitudinal sectional SEM microstructures of the in situ composites after cold deformation with η = 4: (a) as-cast Cu-7Cr alloy, (b) DS Cu-7Cr alloy, (c) as-cast Cu-7Cr-0.1Ag alloy, (d) DS Cu-7Cr-0.1Ag alloy

3.2 Mechanical properties

Fig. 3 presents the tensile strength of the deformation-processed Cu-7Cr-0.1Ag in situ composites with a cumulative cold deformation strain of $\eta = 7$ from the as-cast alloy and the DS alloy with a growth rate of 200 $\mu m \cdot s^{-1}$. As shown in Fig. 3, before the 500℃×1h intermediate heat treatment, the tensile strength of the in situ composite from the DS alloy is higher than that from the as-cast alloy. This is attributed to the fact that the Cr fiber-like grains formed in the DS Cu-7Cr-0.1Ag alloys parallel to the pulling direction and are finer and more uniform compared to that of the as-cast alloy, leading to improved tensile strength. The intermediate heat treatment decreases the tensile strength of both the DS and as-cast Cu-7Cr-0.1Ag in situ composite. This is mainly attributed to the coarsening of Cr fibers, the recovery and recrystallization of Cu matrix

and the conglomeration of precipitated Cr particles. After the 500℃×1h intermediate heat treatment, the tensile strength of the Cu-7Cr-0.1Ag in situ composite from the DS alloy is higher than that from the as-cast alloy, while the tensile strength gap of the Cu-7Cr-0.1Ag in situ composite from DS alloy produced by the intermediate heat treatment is lightly greater than that from the as-cast alloy. DS technique promotes the formation of finer and more uniform Cr fiber-like grains parallel to the pulling direction as previously mentioned, which increases the tensile strength. The finer Cr fibers of the DS Cu-7Cr-0.1Ag in situ composite would cause a great decrease in the tensile strength by reducing the heat stability during heat treatment [27]. So the tensile strength change of the in situ composite from the DS alloy produced by the intermediate heat treatment is more significant than that from the as-cast alloy. Similar results are obtained when the DS technique and the intermediate heat treatment process are employed to produce the deformation-processed Cu-7Cr in situ composites.

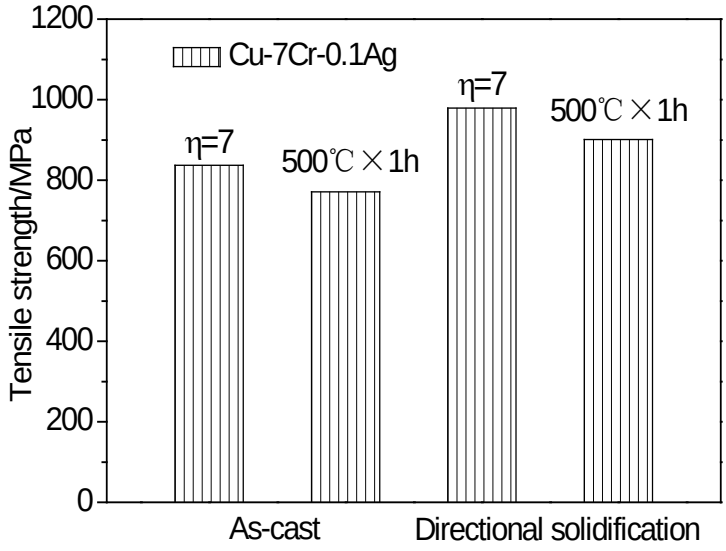

Fig.3 Tensile strength of the deformation-processed Cu-7Cr-0.1Ag in situ composites at η = 7 from the as-cast alloy and the DS alloy with a growth rate of 200 μm·s^{-1} before and after 500℃×1h intermediate heat treatment

Fig. 4 presents the tensile strength of the deformation-processed Cu-7Cr-0.1Ag in situ composites with a cumulative cold deformation strain of η = 8 after isochronic aging treatment for 1 h produced by the thermo-mechanical treatment from the as-cast alloy and the DS alloy with a growth rate of 200 μm·s^{-1}. Fig. 4 shows that the tensile strength of both the in situ composites

from the as-cast and DS alloys increases gradually after the isochronal aging treatment for 1 h. The tensile strength reaches a peak value at 400 ℃, then it is progressively lower after isochronal aging treatments at higher temperatures. The aging treatment below 400 ℃ accelerates Cr precipitation and does not or seldom induce recovery and recrystallization of the Cu matrix, which increases the tensile strength. The aging treatment above 400 ℃ promotes the coarsening of Cr fibers, the conglomeration of precipitated Cr particles, the recovery and recrystallization of the Cu matrix, which are factors decreasing the tensile strength. Similar to the effect of different intermediate heat treatments on tensile strength, the tensile strength of Cu-7Cr-0.1Ag in situ composite from the DS alloy with a growth rate of 200 $\mu m \cdot s^{-1}$ is higher than that from the as-cast alloy under the same aging temperature. In addition, the tensile strength gap lowers than 400 ℃ is greater than that beyond 400 ℃. This is because that the DS technique refines the Cr phase, which induces the tensile strength peak of the Cu-7Cr-0.1Ag from the DS alloy to shift to a lower temperature by reducing the heat stability of the Cr fibers during the aging treatment [27]. Similar results are obtained when the isochronic aging treatment process is employed to the deformation-processed Cu-7Cr in situ composites with a cumulative cold deformation strain of η = 8 from the as-cast alloy and the DS alloy with a growth rate of 200 $\mu m \cdot s^{-1}$.

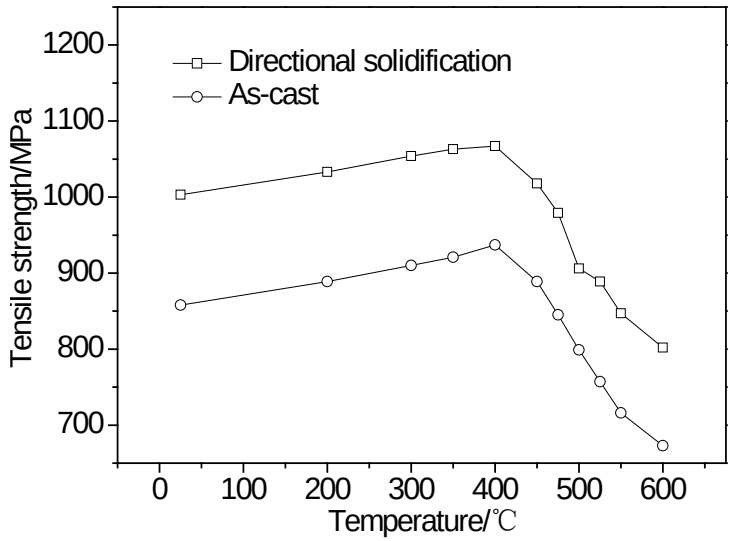

Fig.4 Tensile strength change curves of the deformation-processed Cu-7Cr-0.1Ag in situ composites at η = 8 from the as-cast and DS alloys after isochronic aging treatment for 1h

Fig. 5 presents the elongation to fracture of the deformation-processed Cu-7Cr-0.1Ag in situ

composite with a cumulative cold deformation strain of η = 8 produced by the thermo-mechanical treatment from the DS alloy with a growth rate of 200 μm·s^{-1} after isochronic aging treatment for 1 h. Fig. 5 shows that the elongation to fracture increases gradually with increasing the temperature of isochronic aging treatment. The elongation to fracture change of the in situ composite at η = 8 from the as-cast alloy after isochronic aging treatment for 1 h shows similar characterization.

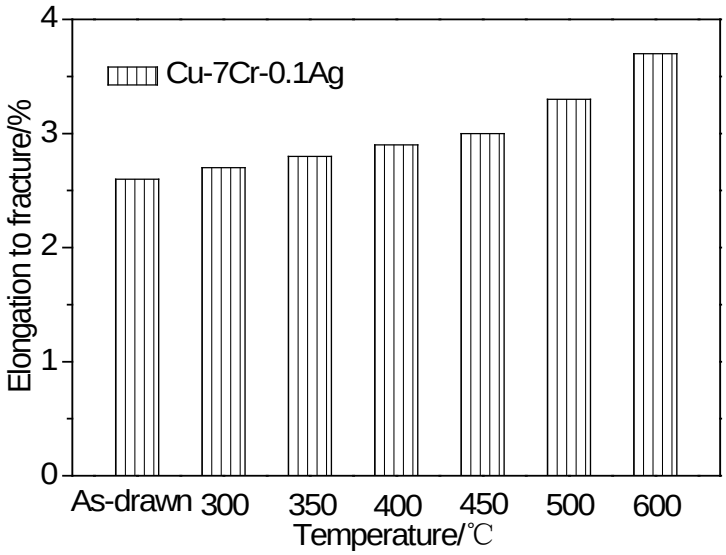

Fig.5 Elongation to fracture of the deformation-processed DS Cu-7Cr-0.1Ag in situ composite at η = 8 after isochronic aging treatment for 1h

3.3 Conductivity

Fig. 6 presents the conductivity of the deformation-processed Cu-7Cr-0.1Ag in situ composites with a cumulative cold deformation strain of η = 7 from the as-cast alloy and the DS alloy with a growth rate of 200 μm·s^{-1}. As shown in Fig. 6, before the 500℃×1h intermediate heat treatment, the conductivity of the in situ composite from the DS alloy is slightly lower than that from the as-cast alloy. There are two main reasons as follows. On the one hand, the Cr fibers of the in situ composite from the DS Cu-7Cr-0.1Ag alloy are finer than that from the as-cast alloy, which decreases the conductivity by increasing the interfacial density. On the other hand, the cooling rate of the DS Cu-7Cr-0.1Ag alloy with a growth rate of 200 μm·s^{-1} is higher than that of the as-cast alloy during solidification, which decreases the conductivity by increasing the supersaturated solid solution of Cr atoms in the Cu matrix [9, 27, 28]. After the intermediate heat treatment of 500℃×1h, the conductivity of both in situ composites from the DS and as-cast alloys increases.

This may be attributed to the fact that the intermediate heat treatment promotes Cr precipitation from the Cu matrix. Previous research [29-31] showed that an adequate intermediate heat treatment led to Ag precipitations, which could optimize the strength and conductivity of binary Cu-based in situ composites. However the Ag contents of the in situ composites in these publications were higher than 6%. The ternary in situ composites consisted of Cu matrix, a non-soluble body-centered cubic refractory metal such as Nb, Fe or Cr, and face-centered cubic Ag, Cu-Ag eutectic [31]. In this paper, the Ag content was low and the Ag content of these two kinds of Ag-containing in situ composites is same. The effect of Ag micro-alloying on the properties of the Cu-7Cr in situ composites during heat treatments is still not clear at present and needs more in-depth investigation. In addition, recovery and recrystallization during the intermediate heat treatment reduce the resistivity [32-34]. The conductivity of the Cu-7Cr-0.1Ag in situ composite from the DS alloy is slightly lower than that from the as-cast alloy after the intermediate heat treatment, while the conductivity gap of the Cu-7Cr-0.1Ag in situ composite from the DS alloy produced by the intermediate heat treatment is slightly greater than that from the as-cast alloy. As previously explained, the DS influences the conductivity of the in situ composites from two aspects. The relatively high supersaturated solid solution of Cr atoms in the Cu matrix increases the precipitation driving force of Cr, improving the conductivity during the intermediate heat treatment [9, 27]. The relatively fine Cr fibers of the in situ composite reduce the heat stability, which decreases the conductivity during the intermediate heat treatment. The increment caused by the former is bigger than the decrement produced by the latter. Therefore, the conductivity increment of the Cu-7Cr-0.1Ag in situ composite from DS is higher than that from the as-cast alloy. Similar results are obtained when the DS technique and the intermediate heat treatment process are employed to produce the deformation-processed Cu-7Cr in situ composites.

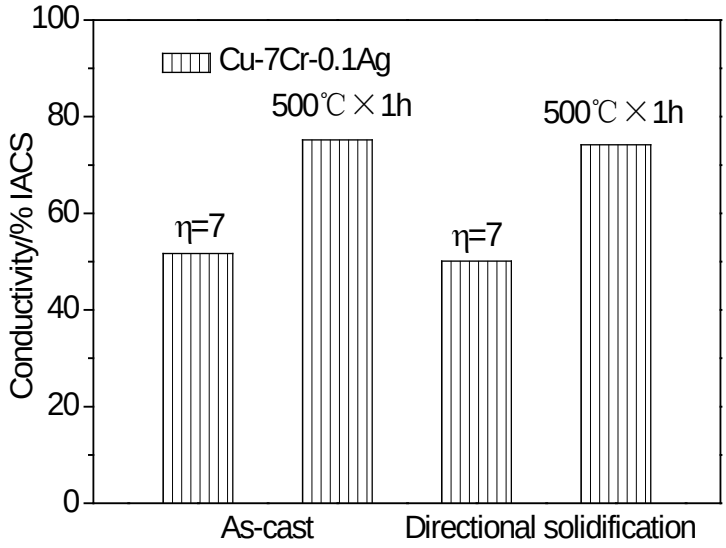

Fig.6 Conductivity of the deformation-processed Cu-7Cr-0.1Ag in situ composites at η = 7 from the as-cast alloy and the DS alloy with a growth rate of 200 μm·s^{-1} before and after 500℃×1h intermediate heat treatment

Fig. 7 presents the conductivity of the deformation-processed Cu-7Cr-0.1Ag in situ composites with a cumulative cold deformation strain of η = 8 after isochronic aging treatment for 1 h produced by the thermo-mechanical treatment from the as-cast alloy and the DS alloy with a growth rate of 200 μm·s^{-1}. Fig. 7 indicates that the conductivity of both the in situ composites from the as-cast and DS alloys increases after the isochronal aging treatment. The increment rises with increasing the aging temperature, which is different from the change of the tensile strength. This is attributed to the fact that the highest aging temperature applied is equivalent to or lower than the temperature of the conductivity peak. Jin [8] investigated the conductivity of a Cu-15 wt% Cr in situ composite by using heat treatment technique. The results showed that the conductivity of the Cu-15 wt% Cr in situ composite increased gradually after the isochronal aging treatment for 1h and reached a peak value at 600℃, and then the conductivity was progressively lower after isochronal aging treatments at higher temperatures. Similar to the influence of the intermediate heat treatment on conductivity, the conductivity of the Cu-7Cr-0.1Ag in situ composite from the DS alloy is slightly lower than that from the as-cast alloy after aging treatment under the same temperature.

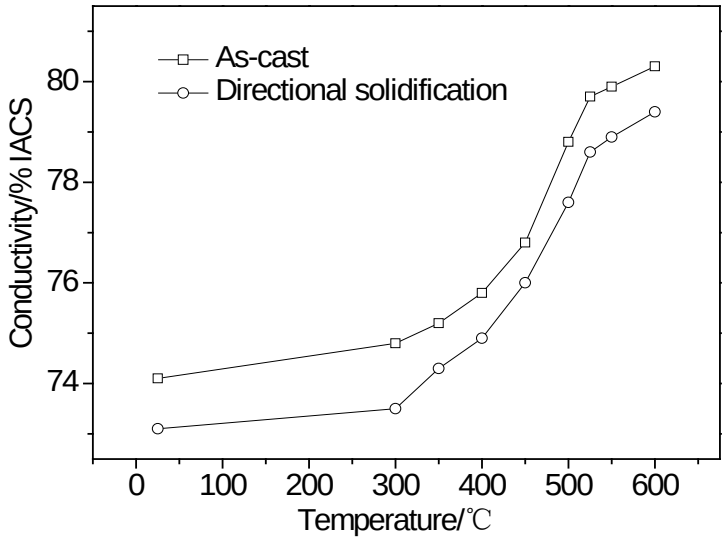

Fig.7 Conductivity change curves of the deformation-processed Cu-7Cr-0.1Ag in situ composites at η = 8 from the as-cast and DS alloys after isochronic aging treatment for 1h

The results obtained from Figs. 4, 5 and 7 indicate that the properties of the deformation-processed Cu-7Cr-0.1Ag in situ composite at η = 8 have been improved by using an appropriate aging treatment. The tensile strength, elongation and conductivity of the deformation-processed Cu-7Cr-0.1Ag in situ composites from the DS alloy with a growth rate of 200 $\mu m \cdot s^{-1}$ reach 1067 MPa, 2.9% and 74.9% IACS; 1018 MPa, 3.0% and 76.0% IACS; 906 MPa, 3.3% and 77.6% IACS, respectively, at η = 8 after isochronic aging treatment for 1h. The results suggest that the tensile strength, elongation and conductivity of the Cu-7Cr-0.1Ag in situ composite could be adjusted and controlled by using DS technique and appropriate aging treatment.

4 Conclusions

(1) The size, shape and distribution of the second-phase Cr grains are significantly different in the Cu-7Cr-0.1Ag alloys with different growth rates. With the increase of the growth rate, the size is smaller and the shape gradually transforms into the rod-like.

(2) At a growth rate of 200 $\mu m \cdot s^{-1}$, the Cr grains in the DS Cu-7Cr-0.1Ag alloy transform into fine and curled Cr fiber-like grains parallel to the pulling direction. The deformation-processed Cu-7Cr-0.1Ag in situ composites from the DS alloys have finer and

more uniform Cr fibers than that from the as-cast alloy at the same cumulative cold deformation strain.

(3) The tensile strength of the deformation-processed Cu-7Cr-0.1Ag in situ composite at $\eta = 7$ from the DS alloy with a growth rate of 200 $\mu m \cdot s^{-1}$ is significantly higher than that from the as-cast alloy, while the conductivity of the in situ composite from the DS alloy is slightly lower than that from the as-cast alloy.

(4) After the intermediate heat treatment, the conductivity increment of the deformation-processed Cu-7Cr-0.1Ag in situ composite from the DS alloy is higher than that from the as-cast alloy. The tensile strength decrement of the in situ composite from the DS alloy is lightly greater than that from the as-cast alloy, but the tensile strength of the in situ composite from the DS alloy is still significantly higher than that from the as-cast alloy.

(5) The tensile strength, elongation to fracture and conductivity of the deformation-processed Cu-7Cr-0.1Ag in situ composites from the DS alloy with a growth rate of 200 $\mu m \cdot s^{-1}$ reach 1067 MPa, 2.9% and 74.9% IACS; 1018 MPa, 3.0% and 76.0% IACS; 906 MPa, 3.3% and 77.6% IACS, respectively, at $\eta = 8$ after isochronic aging treatment for 1h.

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