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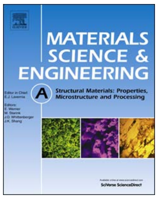

Effects of annealing treatment on the microstructure evolution and the strength degradation behavior of the commercially pure Al conductor

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ABSTRACT

Strength degradation during the electron transmission process is always a hidden danger to the overhead transmission lines. In this study, the microstructure evolutions and the strength degradation behaviors of a cold-drawn commercially pure Al conductor (CPAC) were investigated systematically by a series of annealing experiments. The results show that the texture evolution, dislocation recovery and subgrain growth during the recrystallization should be responsible for the strength degradation of CPAC. Besides, the microstructure evolution depends on the annealing temperature. For instance, some of the < 111 > texture was changed into the < 001 > one in the CPACs annealed at a temperature of 90 °C; while, there is an obvious increase in the subgrain width when the CPACs were annealed in the high temperature range from 150 °C to 300 °C. Finally, the strength degradation due to the texture evolution and the subgrain coarsening was quantitatively calculated based on the crystallographic analysis.

1. Introduction

Commercially pure Al (CPA) and Al-Mg-Si alloy are widely used as conductive material in power industry owing to their relatively good electrical conductivity, high strength-to-density ratio, outstanding corrosion resistance and lower cost compared with other conductive materials, e.g., Cu [1–3]. As a conductive material, CPA is widely applied as the external-layer of the Al conductor steel reinforced (ACSR), which is a kind of overhead power transmission lines. Besides, all Al alloy conductor (AAAC), another kind of overhead power transmission lines, is wholly made of Al-Mg-Si alloy [4,5]. In particular, strength is a very important property of the Al conductors suffering various loads caused by wind, ice and the weight of the conductor when they are in service [6,7]. With the increasing demands of electricity in the past years, there has been an intense desire for the higher transmission capacity, which can lead to a higher operating temperature for the Al conductors [8]. However, the growths of grain and second phase in Al conductor often occur as the operating temperature increases over the critical value, which results in strength degradation. For example, the operating temperature of the Al-Mg-Si wire is limited to be lower than 90 °C, as the growth of precipitates can occur if the temperature is over the critical value [6,9]. As a result, the thermal-resistant properties of the Al conductors determine the operating temperature. In addition, the recovery of dislocation, the growth of grain and the transformation of precipitate have been observed during the annealing process which usually leads to the strength degradation of the deformed metals [10–14]. The strength loss in the Al alloy during the annealing process can be subdued by adding Zr and Sc, as the thermo-stable Al_3Zr and Al_3Sc can efficiently pin grain boundaries to suppress the grain growth [8,15,16]. Therefore, the operating temperature of Al-Mg-Si alloy can be improved to 180 °C by adding Zr [16]. Besides, the strength of Al alloy conductors including Al-Zr and Al-Sc alloys still maintains stable at elevated temperature as high as 230 °C [9]. Although the strength degradation behaviors and the microstructure evolution of the Al alloy conductors have been widely investigated [8,9,15–18], the strength degradation behavior of the annealed CPACs is rarely reported [6]. In the present work, the strength degradation behaviors of the CPACs annealed at various temperatures were estimated to acquire their safety operating temperature. Additionally, the microstructure evolution in the annealed CPACs was examined using electron backscatter diffraction (EBSD) and transmission electron microscopy (TEM) to quantitatively reveal the mechanisms behind the strength degradation as well as to present the rules for designing the thermal-resistant Al alloy conductor.

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2. Materials and methods

The chemical composition of the CPAC (wt%) adopted in this study is: Al > 99.6, impurity ≤ 0.4. The CPAC with a diameter of 2.98 mm was manufactured by cold drawing on a bull block drawing machine for 9 passes from the original aluminum rod with a diameter of 9.50 mm. The total area reduction is ~90.2%. The cold-drawn CPACs were annealed at 90 °C, 150 °C, 200 °C, 250 °C and 300 °C for various holding time in a thermotank and then cooled by water. The uniaxial tensile tests of the annealed CPACs with a gauge length of 150 mm were carried out on an Instron 5982 testing machine. Tensile tests were performed at room temperature with a constant strain rate of $1.0 \times 10^{-3}\,s^{-1}$ with the tensile axis being parallel to the cold-drawing direction. Some cylindrical Al pieces and rectangular Al pieces cut from the radial section and the axial section of the CPACs were respectively fabricated into specimens for the EBSD observations. Metallographic samples were polished using 2000# SiC paper and then electrolytic polished for ~90 s at 0 °C using an etching solution containing 10% perchloric acid and 90% alcohol by volume. Subgrain size distributions and orientation distribution maps were measured and analyzed by EBSD technique integrated into ZEISS SUPRA 35 scanning electron microscope (SEM). For the TEM observations, samples were cut from the radial sections and the axial sections of the CPACs respectively, ground to a thickness of ~0.05 mm and then twin-jet electro-polished at − 20 °C using a solution of 20% perchloric acid and 80% methanol by volume. The TEM foils were examined using an FEI Tecnai F20 microscope operating at 200 kV.

3. Results

3.1. *Strength degradation behavior of the CPACs during the annealing treatments*

The relations between the ultimate tensile strength (UTS) as well as the yield strength (YS) and the holding time of the CPACs annealed at different temperatures are shown in Fig. 1. Clearly, the YS of the CPAC annealed at 90 °C for different holding time is finally stable at ~194.0 MPa, which is slightly lower than the YS of the cold-drawn CPACs (198.8 MPa). However, with the increase of annealing temperature to 150 °C, 200 °C and 250 °C for enough holding time, the YSs of the annealed CPACs finally decrease to ~168.1 MPa, ~142.5 MPa and 91.4 MPa, respectively. For the CPACs annealed at 300 °C, their YSs continuously decrease down to 45.5 MPa with increasing the holding time to 12 h.

3.2. *TEM observations of the annealed CPACs*

The CPACs annealed at 90 °C for 72 h, 150 °C for 48 h, 200 °C for 4 h, 250 °C for 2 h and 300 °C for 20 min were observed by TEM to show their microstructure evolution, as their YSs continuously decrease. The microstructures observed from the radial direction by TEM are shown in Fig. 2. Fine subgrains with the relatively low-level dislocation densities are the typical characteristics of the cold-drawn CPACs (Fig. 2a). Specifically, the dislocation network could be observed in the subgrain of the CPACs annealed at 90 °C for 72 h (Fig. 2b), providing a convincing evidence for the occurrence of dislocation recovery [19,20]. Besides, the subgrains in the radial section of the annealed CPACs gradually grew with increasing the annealing temperature. For the CPACs annealed at 300 °C for 20 min, an abnormal grain growth was observed, showing relatively inhomogeneous microstructures.

Fig. 3 shows the axial microstructures of the annealed CPACs, in which the subgrains were obviously elongated in the drawing direction (Fig. 3a). It is clear that the subgrains in the CPACs annealed at 90 °C for 72 h still maintain the elongated state (Fig. 3b), indicating a stable subgrain size. However, Fig. 3c-3f show that the lengths of subgrains obviously decrease in the CPACs annealed at the higher annealing temperatures due to the recrystallization of subgrains. Consequently, the subgrains of the CPACs annealed at high temperatures are wider and shorter as compared with those of the cold-drawn CPACs.

3.3. *EBSD observations of the annealed CPACs*

Texture, as one of important microstructures, can significantly influence the mechanical properties of Al wires. It was reported that texture is considered as a kind of strengthening factor for the metallic materials fabricated by the cold-drawing process and the SPD technologies [7,21–23]. In addition, the major < 111 > texture and the minor < 001 > texture were the typical texture components observed in the heavily cold-drawn FCC metals including Ag, Cu and Al [24,25]. Furthermore, the ratio of < 111 > to < 001 > texture varies with the drawing strain, stacking fault energy and drawing temperature [24–26]. Texture evolution of the annealed CPACs is given in Fig. 4. The orientation distribution map of the cold-drawn CPAC observed from the radial section shows that the CPACs have a weaker < 001 > fiber texture together with a stronger < 111 > fiber texture. However, an increasing amount of subgrains with < 001 > orientation and a decreasing amount of subgrains with < 111 > orientation were found in the annealed CPACs, indicating the occurrence of texture evolution from the < 111 > to the < 001 > texture. The volume fraction of the < 111 > texture was calculated by the following equation:

$$f_{<;111>} = \frac{S_{<111>}}{S_0}, \tag{1}$$

where $f_{<111>}$ is the volume fraction of the < 111 > texture; $S_{<111>}$ is the area of the subgrains with < 111 > orientation and S_0 is the measured total area.

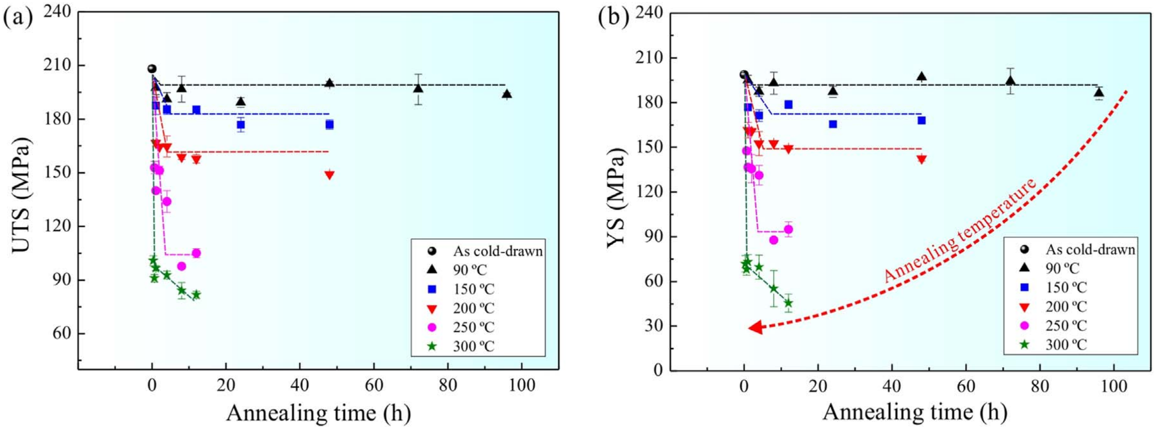

Fig. 1. The relations between (a) the ultimate tensile strength (UTS) as well as (b) the yield strength (YS) and the holding time for the CPACs annealed at different temperatures.