INTRODUCTION

Low-alloy Cu–Cr copper-based alloys are used in various industries as electrical materials [1–3]. However, a known problem for these alloys is to enhance their strength even more while retaining high electric conductivity; this problem can be solved by the refinement of the grain structure.

In recent years, the refinement of the grain structure in various metals and alloys has been implemented using the methods of severe plastic deformation (SPD), which were based on the use of severe shear deformations under high pressures and at low homologous temperatures [4, 5].

Dispersed particles that precipitate after quenching and aging also play an important role in the strengthening of Cu–Cr copper alloys [6]. In these alloys, aging is carried out at temperatures that exceed substantially the temperature of the onset of intense grain growth in UFG pure copper. Because of this, the processes of grain growth can also occur in the UFG state during aging, which leads to the softening of the material. The precipitation of dispersed particles in an UFG alloy can favor both the attainment of a high-strength state and the hampering of grain growth during heating.

On the other hand, when a high-strength state is attained in a copper-based alloy, another important problem arises, namely, to retain an increased electric conductivity.

It is known that the electric resistance of metal materials increases when stacking faults, such as atoms of alloying elements, impurities, dislocations, dispersed particles, grain boundaries, etc., are present in a material [7–12]. For example, the alloying of copper with chromium in a concentration of 0.4–1.0 wt % (the BrKh alloy) leads to a decrease in the electric conductivity at room temperature by ~16% [1]. The formation of an average grain size of 150 nm in pure copper using severe plastic deformation by torsion (SPDT) favors an increase in the electric conductivity by less than 5% [11]. For comparison, the contribution of dislocations to the electric conductivity at a density of dislocations of \(10^{11}\) cm\(^{-2}\) is \(2.3 \times 10^{-10}\) Ohm m, which is ~1.3% of the electric conductivity of pure copper [13, 14]. At a concentration of vacancies of \(10^{-5}\), their contribution to the electric conductivity is \(2 \times 10^{-10}\) Ohm m, which is less than 1.2% of the electric conductivity of pure copper [13, 14].

It follows from the above-presented estimates that in the UFG Cu–Cr copper-based alloy, the contribution of the alloying elements to the electric resistance at room temperature will substantially exceed the contribution of grain boundaries, dislocations, and vacancies. Therefore, a decrease in the concentration of the alloying elements in the lattice inside the grains via their redistribution to the grain boundaries would favor the retention of the electric conductivity.
Table 1. Chemical composition of the copper-based alloy

<table>
<thead>
<tr>
<th>Element</th>
<th>Cr</th>
<th>Ag</th>
<th>Fe</th>
<th>P</th>
<th>Si</th>
<th>Cu</th>
</tr>
</thead>
<tbody>
<tr>
<td>Wt %</td>
<td>0.5</td>
<td>0.12</td>
<td>0.06</td>
<td>0.06</td>
<td>0.05</td>
<td>Balance</td>
</tr>
</tbody>
</table>

One of the possible ways to enhance the strength, while retaining the electric conductivity, is to produce a UFG structure in which atoms of the alloying elements would be localized as dispersed particles of the precipitations located over the grain boundaries and in the triple junctions; in this case, the density of dislocations would be low. This structure can be produced via the development of dynamic aging [15], since the precipitation of the particles is a diffusion-induced process and in UFG materials the activation energy of grain-boundary diffusion decreases [16].

Thus, attaining both high strength and high electric conductivity in the Cu–Cr copper-based alloys requires the development of optimum conditions of severe plastic deformation and subsequent heat treatment, which is the aim of this work.

EXPERIMENTAL

As material for the study, we selected the Cu–Cr copper-based alloy produced by Wieland Co. (Germany), which, according to the results of optical-emission analysis carried out using a Bruker Q4 TASMAN spectrometer, had the chemical composition presented in Table 1.

Before quenching, a protective coating was deposited on the surfaces of the original copper-based alloy specimens to prevent the surfaces against oxidation. The specimens were then subjected to heating in air for 2 h to a temperature of 1050°C and quenching in water.

The UFG structure was produced in specimens 20 mm in diameter and 1 mm thick using a setup for severe plastic deformation by torsion in the temperature range of 20–450°C under a pressure of 6 GPa; the number of revolutions was equal to 10.

Equal-channel angular pressing (ECAP) was carried out in eight passes following the Bc route using a tool with an angle of intersection of the channels of 90° at temperatures of 20 and 300°C; specimens 10 mm in diameter and 60 mm long were used.

The structural examinations were performed using a JEM-6390 scanning electron microscope and a JEM-2100 transmission electron microscope. Thin foils were prepared in a Tenupol-5 apparatus by jet electrolytic polishing at a voltage of 22–24 V using electrolyte of the following composition: 920 mL of water (Н₂О), 70 mL of orthophosphoric acid (Н₃РО₄), and 15 mL of glycerol (C₃H₅(OH)₃). The structure of the ECAP specimens was examined in the cross section. The structure of the SPDT specimens was examined in the middle of the radius.

The Vickers microhardness was measured using a Micromet 5101 apparatus along the diameter of the SPDT specimens and in the cross section of the ECAP specimens for 10 s under a load of 100 g.

The specific electric conductivity was measured at room temperature by the eddy-current method using a VE-27NTs apparatus with a relative measurement error of 2%. It is known that, according to the International Standard IASC (International Annealed Copper Standard), the notation 100% IACS corresponds to annealed pure copper with an electric conductivity of 58 MS/m (the electric resistance is 0.017241 μOhm m) [1]. Because of this, in this work, the results of measuring the electric conductivity are presented in % IASC, i.e., the percentage of the electric conductivity of pure copper.

The unidirectional tensile tests were carried out using a test rig for the deformation of small specimens at room temperature at a rate of 3 × 10⁻³ s⁻¹. In each state, two specimens with 4.0 × 1.0 × 0.7-mm gage portions were tested; the specimens were cut out in the longitudinal section of the ECAP blanks. The gage portion of the SPDT specimens passed through the middle of the radius.

RESULTS

The microstructure of the alloy in the original state was characterized by an average grain size of ~37 μm and contained inclusions of a fine strengthening phase.

After quenching and SPDT at a temperature of 20°C, the structure of the copper-based alloy was refined to an average size of structural elements (grains, subgrains, and cells) of ~200 nm (Fig. 1). Complex diffraction contrast in the electron micrographs is indicative of the presence of high internal stresses in the structure of the specimens, which resulted from the effect of severe shear deformations under high pressures. An analysis of the bright-field and dark-field images of the structure of the UFG specimens produced by SPDT at a temperature of 20°C has not revealed particles of precipitates in the specimens. Additional 30-min annealing of these specimens at temperatures of 300 and 400°C has not led to substantial changes in the average size of the structural elements and to the appearance of visible particles of precipitates.

After a 30-min annealing at a temperature of 500°C, in the UFG specimens produced by SPDT at a temperature of 20°C, 10-nm particles of precipitates are observed against the background of the grains with