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Effect of Alloying with Palladium on the Electrical and Mechanical Properties of Copper

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Abstract—Structure and physicomechanical properties of Cu–Pd alloys that contain 0.5–5.9 at % Pd have been studied. It has been shown that, in all alloys, a solid solution is formed; the lattice parameter of the fcc lattice and the electrical resistivity of the alloys grow linearly with an increase in the content of palladium. It has also been revealed that the introduction of palladium leads to an increase in the recrystallization temperature and to an increase in the strength properties. The assumption on the formation of an atomic short-range order in the quenched Cu–4.6 at %Pd and Cu–5.9 at %Pd alloys has been made.

Keywords: copper–palladium alloys, phase diagram, mechanical properties, resistivity measurements

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INTRODUCTION

The miniaturization of electronic and electric devices forces designers to search for various ways to reduce the sizes and weight of all components. The use of low currents in modern devices makes it possible to decrease the cross-section of wires. However, in this case, it becomes necessary to guarantee their mechanical integrity upon overloads during operation. Therefore, the search for methods of strengthening copper is actively being conducted by many research groups both in Russia and abroad.

To solve this problem, different alloying methods are used. For example, the addition of even a small amount of magnesium to copper leads to a sharp increase in its strength properties due to solid-solution strengthening [1, 2]. The Cu–Ag alloys can be strengthened as a result of precipitation hardening [3]. The creation of copper–niobium composites based on the methods of hydroextrusion [4] and works on strengthening a copper matrix by introducing nanodispersed particles of different metals (e.g., chromium) using the methods of severe plastic deformation (SPD) [5] can also be mentioned.

As a rule, an increase in the strength properties simultaneously causes a reduction in some other functional characteristics. For example, the corrosion resistance of Cu–Mg alloys is lower than that of pure copper [2]. The introduction of a large number of niobium fibers into the copper matrix leads to an undesirable increase in the density of the conductor. The nanosized particles are centers of electron scattering; therefore, precipitation-hardened alloys have a reduced electrical conductivity. The use of severe plastic deformation (SPD) efficiently solves the problem of increasing the strength properties of a material due to the refinement of its grain structure, but simultaneously leads to a sharp reduction in its plasticity and the thermal stability of its properties. Furthermore, most SPD procedures either require large energy inputs [6] or make it possible to work only with small samples [5]; therefore, these experiments are only carried out on a laboratory scale and are of purely scientific interest.

Recently, additional requirements concerning corrosion resistance are presented to conductors. To solve this problem, copper conductors are covered with a thin layer of palladium [7]. However, in the process of exploiting these wires, palladium diffuses into copper fairly rapidly, forming various ordered phases at the interface.

At the same time, it is known that the introduction of even small amounts of palladium significantly increases the corrosion resistance of copper [8]. As follows from the phase diagram (Fig. 1), at a palladium content of less than 10 at %, a continuous series of solid solutions is formed in the Cu–Pd system; an increase in the palladium concentration leads to the formation of superstructures ordered according to the $L1_2$ or $B2$ type at the compositions close to Cu_3Pd and CuPd, respectively [9]. In addition, it can be assumed that the introduction of palladium into copper will

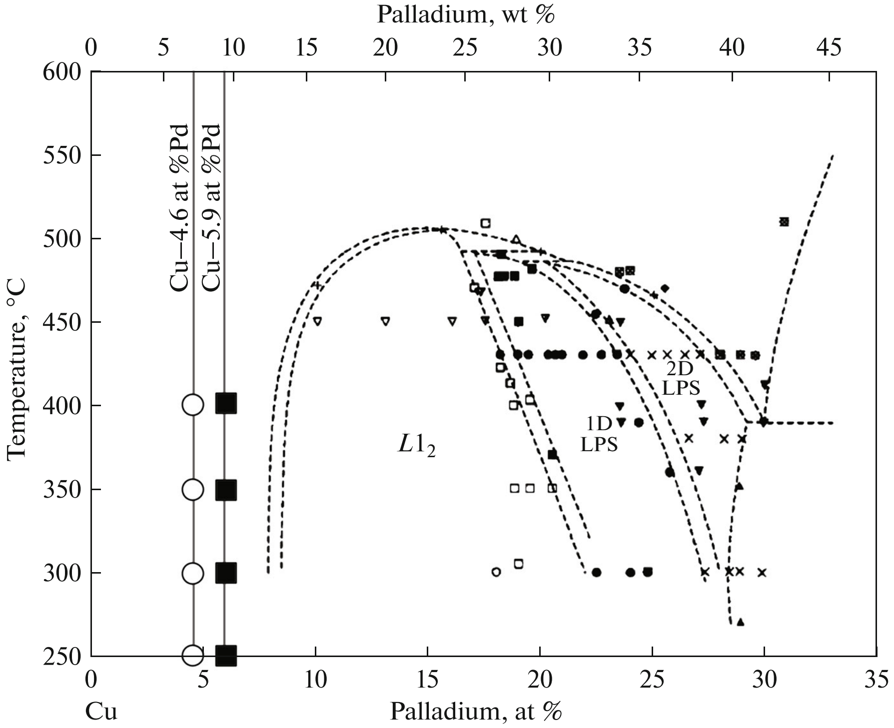

Fig. 1. Portion of the phase diagram of Cu–Pd alloys [9]. Vertical lines show compositions of the studied alloys. Symbols (○) and (■) designate the treatment temperatures.

lead to grain refinement and to an increase in the temperature of the recrystallization of the alloys, similar to what was observed previously in [10] upon the addition of rhenium. Thus, it is of interest to investigate the physicomechanical properties of Cu–Pd alloys with a low content of palladium from the viewpoint of the opportunity of using them as conducting materials.

An analysis of the literature data indicates that the microstructure and physicomechanical properties of these alloys have been insufficiently investigated. For example, the concentration dependences of the electrical resistivity of the Cu–Pd alloys were constructed ~80 years ago [11, 12], but the alloys with the palladium content of less than 10 at % have not been investigated in those works.

Table 1. Composition of Cu–Pd alloys

Alloy	Content of Pd, wt %	Content of Pd, at %
1	0.8	0.5
2	1.5	0.9
3	2.4	1.4
4	5.0	3.0
5	7.5	4.6
6	9.5	5.9

This work is aimed at the study of the influence of small additions of palladium on the mechanical and electrical properties of copper.

EXPERIMENTAL

The Cu–Pd alloys were prepared using copper and palladium with a purity of 99.98% in a vacuum no worse than 10^{-2} Pa by the method of double remelting with the casting into a graphite crucible. The compositions of the alloys, refined according to the results of X-ray microprobe analysis (Superprobe JCXA-733, accelerating voltage 25 kV, the probe current 50 nA), are given in Table 1. Below, the composition of the alloys will be indicated in at %.

Ingots with a diameter of 8 mm were homogenized for 3 h at a temperature of 850°C with the subsequent rapid cooling in water. Then, ingots were subjected to mechanical and thermal treatment including several sequential operations, in particular, deformation by rolling to 90%, annealing at a temperature of 700°C with a holding for 1 h, and water quenching. The samples in two initial states were investigated: after quenching from 700°C and after preliminary deformation. The mechanical tests were performed using wire samples with a diameter of 1.5 mm obtained by drawing to 90% ($e \approx 2.3$). The X-ray diffraction analysis was carried out using plates 0.2 mm thick, which had been deformed by rolling to $e \approx 3.8$. The greatest true deformation ($e \approx 7.1$) was achieved by drawing the initial ingot into a thin wire with a diameter of 0.22 mm, which was used for resistivity measurements. Furthermore, to compare the microstructure and mechanical properties, we also used samples prepared from the rolled copper wire 8 mm in diameter produced at ZAO SP KATUR-INVEST. The rolled wire was obtained by the method of continuous casting and rolling (Contirod) from cathodic copper of grade M00k (Russian Standard GOST 546-2010). For the correct comparison of the evolution of the structure and properties of pure copper and copper–palladium alloys, they were subjected to identical thermomechanical treatments.

The X-ray diffraction analysis was performed using a DMAX 2200 (Rigaku) diffractometer. The method of continuous surveying at a rate of 4 deg/min was employed. The CuK_α radiation was monochromated by a graphite single crystal.

The resistivity ρ of the samples was measured using the standard four-probe method (with a dc current $I = 20$ mA); the equipment and the procedure were described in detail in [13]. The measurements of the resistivity of the samples at room temperature were conducted on a special conductor, as was described earlier in [14]. The temperature dependences of the resistivity were obtained by heating the samples at a rate of 120 K/h. All heat treatments were carried out in evacuated quartz or glass ampoules.

The mechanical tests were carried out using a ZD 10/90 tensile testing machine at a rate of tension equal to 3 mm/min; the length of the gage part of the samples was 30 mm. For each structural state, no fewer than five samples were tested.

EXPERIMENTAL RESULTS

X-Ray Diffraction Analysis

As was indicated above, the phase composition of the Cu–Pd alloys with a low content of palladium has hardly been studied before. Let us consider, for example, the corresponding part of the Cu–Pd phase diagram presented in Fig. 1 [9]. It can be seen clearly that the experimental points are nearly absent in the portion of the phase diagram with a content of palladium of less than 10 at %. However, recently, additional interest appeared in a more detailed study of precisely these compositions, since it was predicted by the methods of computer simulation [15] that a Cu_7Pd phase can exist at low temperatures. Thus, the low-temperature region of the left-hand side of the phase diagram in Fig. 1 needs refining.

We showed previously that a preliminary SPD of Cu–Pd alloys near the equiatomic composition leads to a considerable acceleration of the $A1 \rightarrow B2$ phase transformation, even at reduced temperatures [14]. A similar conclusion was drawn in [16, 17] upon studying alloys near the stoichiometry of Cu_3Pd, i.e., a preliminary megaplastic deformation causes the giant acceleration of the processes of atomic ordering. Therefore, the phase composition of the alloys in this work was studied after prolonged low-temperature annealings of the samples taken in two initial states: quenched from 700°C; and after preliminary deformation by rolling ($e \approx 3.8$).

Figure 2 displays the X-ray diffraction patterns of the samples of alloy 6 (containing 5.9 at % palladium, Table 1). The diffractogram *1* corresponds to the initial (deformed to $e \approx 3.8$) state of the alloy; the XRD patterns *2–5* were obtained after annealing preliminarily deformed samples at temperatures of 200–350°C (in steps of 50 K). The duration of the annealings was one week (168 h). The lines of only fcc phase are present in all diffractograms; no reflections from other phases have been revealed. With an increase in the treatment temperature, the X-ray diffraction lines become sharper and more perfect; the general shape of the diffractograms changes insignificantly. It is worthwhile to note that a clearly pronounced rolling texture is observed in the XRD pattern of the deformed sample. After annealing the deformed samples in the temperature interval indicated, the preferred orientations of the structure are retained: in all diffractograms, the intensity of peak (220) is an order of magnitude greater than that of other peaks (Fig. 2).

The experiments on the refinement of the phase composition after prolonged annealing were carried out also on the alloy samples containing 4.6 at % Pd. The experimental points obtained are shown for clarity in the phase diagram (Figure. 1). Thus, X-ray diffraction analysis made it possible to establish that, in all of the alloys investigated, a solid solution with an fcc lattice is formed.

It follows from the results obtained that the addition of palladium causes an increase in the lattice parameter of the alloy. For example, the lattice parameter of the well-annealed copper is $a = 0.3615$ nm [9, 18], whereas in an alloy with 5.9 at % Pd, we have $a = 0.3640$ nm. The dependence of the lattice parameter on the content of palladium in the investigated Cu–Pd alloys is shown in Fig. 3. The result obtained confirms the previously drawn conclusion that the dependence of the lattice parameter on the composition of the Cu–Pd alloys (with an fcc structure) is almost linear and corresponds well to the Vegard law [9].

Resistivity and TCR of Alloys

Figure 4 displays the temperature dependences of the electrical resistance that were obtained upon heating at a rate of 120 K/h of the preliminarily deformed samples of the alloys investigated ($e \approx 7.1$). The almost linear character of all the graphs indicates that, in the range of room temperature to 550°C, no phase transformations occur in these alloys. The most noticeable deviation from the linearity is observed in the temperature dependence of resistivity $\rho(T)$ obtained upon

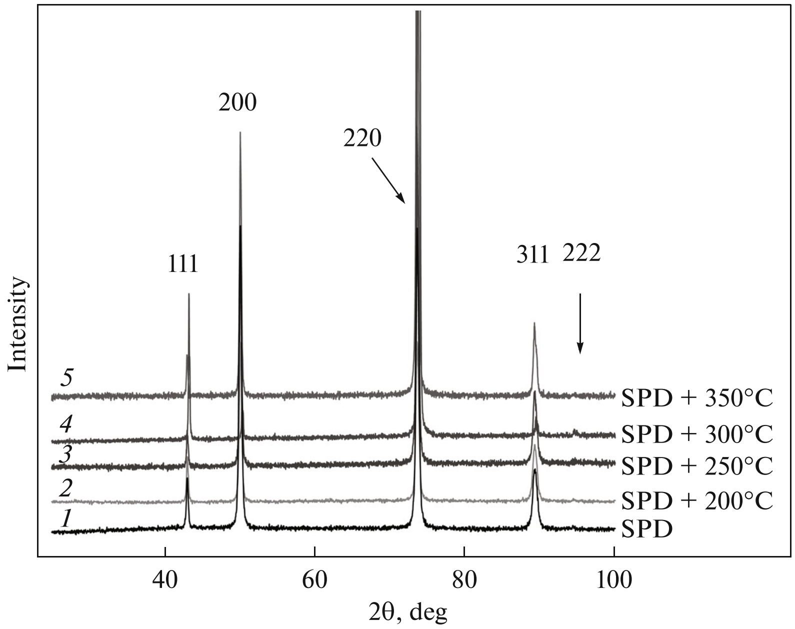

Fig. 2. X-ray diffraction patterns of samples of the Cu–5.9Pd alloy: (*1*) initial deformed state ($e \approx 3.8$); (*2*)–(*5*) states obtained after the annealing of the deformed samples for 168 h in the range of temperatures of 200–350°C.

heating of the sample of the alloy with the maximum content of palladium, Cu–5.9Pd (curve *6* in Fig. 4). Indeed, upon comparison with the temperature dependence of the resistivity of the alloy Cu–4.6Pd (curve *5* in Fig. 4), a small increase in the resistivity near the temperature of 250°C can be seen and, upon further heating of the Cu–5.9Pd alloy, a certain reduction is observed in the rate of change in $\rho(T)$. This deviation from the linear dependence of $\rho(T)$ reflects certain processes of reconstructing the structure of the alloy in the course of heating in order to understand which additional microstructural studies are required. It is necessary to note here that a process such as recrystallization, which necessarily occurs in the preliminarily deformed samples upon their heating, leads only to a small reduction in the electrical resistance [19]. It is impossible to detect the appropriate change in the temperature dependence of resistivity (Fig. 4). To detect the development of the processes of recrystallization, measurements of the resistivity of samples in the course of isothermal annealing are required (this will be considered in more detail below).

Based on the results given in Fig. 4, we constructed the concentration dependences of the resistivity and of the temperature coefficient of resistivity (TCR), which are shown in Figs. 5a and 5b, respectively. The resistivity of the samples of copper in our experiments

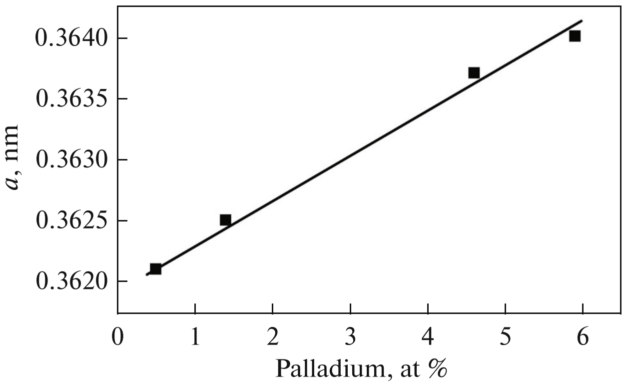

Fig. 3. Dependence of the lattice parameter on the content of palladium in the investigated Cu–Pd alloys.

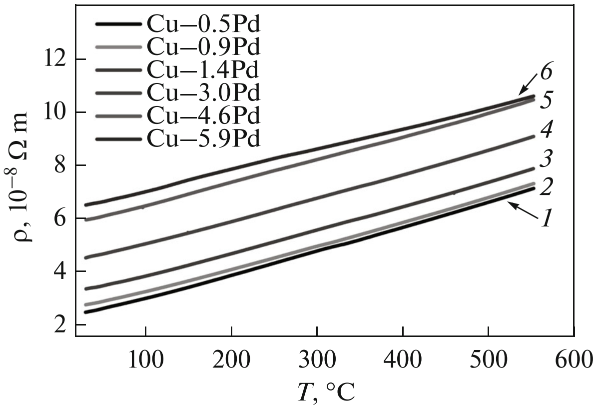

Fig. 4. Temperature dependences of the resistivity of the investigated alloys. Initial state of the samples is the state after deformation by drawing to $e \approx 7.1$. Heating rate was 120 K/h. Order numbers of curves correspond to the order numbers of the alloys in Table 1.

was $\rho = 1.78 \times 10^{-8}\ \Omega$ m. This value somewhat exceeds the tabulated values for annealed copper ($\rho = 1.7241 \times 10^{-8}\ \Omega$ m [5]) and corresponds to the deformed copper of commercial purity. Upon alloying with palladium, the resistivity increases and, at a content of palladium equal to 5.9 at %, it reaches $\rho = 6.35 \times 10^{-8}\ \Omega$ m (Fig. 5a). In the same figure, the literature data on the resistivity of alloys with a higher content of palladium are given for comparison, i.e., for Cu–10Pd, from [11] and, for Cu–11Pd, from [12]. It can easily be seen that all of the values fall into one straight line. The concentration dependence of resistivity plotted in Fig. 5a indirectly confirms the data of X-ray diffraction analysis, which indicates the formation of a solid solution in all investigated Cu–Pd alloys.

In practice, the relative electrical conductivity of a conductor, which is measured according to the International Annealed Copper Standard (IACS), is currently used increasingly more frequently rather than the resistivity. In terms of this standard, the conductivity of any material is indicated in the percent of the electrical conductivity of pure copper. In accordance with the data obtained, the electrical conductivity of the investigated alloys changes from 76% IACS for alloy *1* to 27% IACS for alloy *6*. Note once more that the cathodic copper taken for our study has smaller electrical conductivity compared with the standard, 97% IACS.

Note that, in Fig. 4, the slopes of the obtained $\rho(T)$ dependences decrease with an increase in the content of palladium. As is known, the TCR in the Cu–Pd system is minimum near the equiatomic composition ($\alpha \approx 3.2 \times 10^{-4}\ K^{-1}$ in the alloy Cu–40Pd [20]). However, the variations in TCR in these alloys with small additions of palladium has not been studied earlier; in [8], the first point in the concentration dependence of the TCR corresponds to the alloy Cu–10Pd. Based on the results obtained, it can be assumed that the TCR of the Cu–Pd alloys under consideration changes significantly upon alloying. For example, the TCR of copper is $\alpha \approx 4.15 \times 10^{-3}\ K^{-1}$ [21]. According to our results, alloying copper with only 0.5 at % Pd reduces this value to $\alpha \approx 3.33 \times 10^{-3}\ K^{-1}$ (Fig. 5b). The TCR of Cu–5.9Pd alloy is $\alpha \approx 1.23 \times 10^{-3}\ K^{-1}$.

As is known, the formation of solid solutions is accompanied by an increase in resistivity and a decrease in the TCR in the alloys compared with the initial metal [22]. The results presented in Figs. 4 and 5 correspond completely to this assertion.

The Fine Structure of Alloys

The alloys with the greatest amount of palladium investigated in this work are located near the boundary of existence of the ordered $L1_2$ phase (Fig. 1). Therefore, in these alloys, we can expect the formation of short-range atomic order during the quenching or the formation of a certain number of nuclei of this ordered phase upon prolonged low-temperature annealings. To observe these processes using direct structural methods is difficult. For example, the X-ray diffraction analysis did not reveal any pretransition states (Fig. 2). To detect the so-called fine structure of solid solutions, e.g., short-range atomic order, the initial stages of solid-solution separation, and precipitation hardening, various structurally sensitive methods are used, the measurements of the electrical properties at different stages of the structure formation among them [19]. For example, the short-range atomic order can easily be revealed by an anomalous drop in the electrical resistance in the course of the deformation of the preliminarily quenched samples, which was also shown earlier on the Cu–Pd alloys with higher contents of palladium [23]. In this case, the enhanced electrical resistivity of the quenched alloys is explained by the formation of nanosized clusters of short-range order. In the course of plastic deformation, they are destroyed; as a consequence, the electrical resistivity of samples decreases [24].

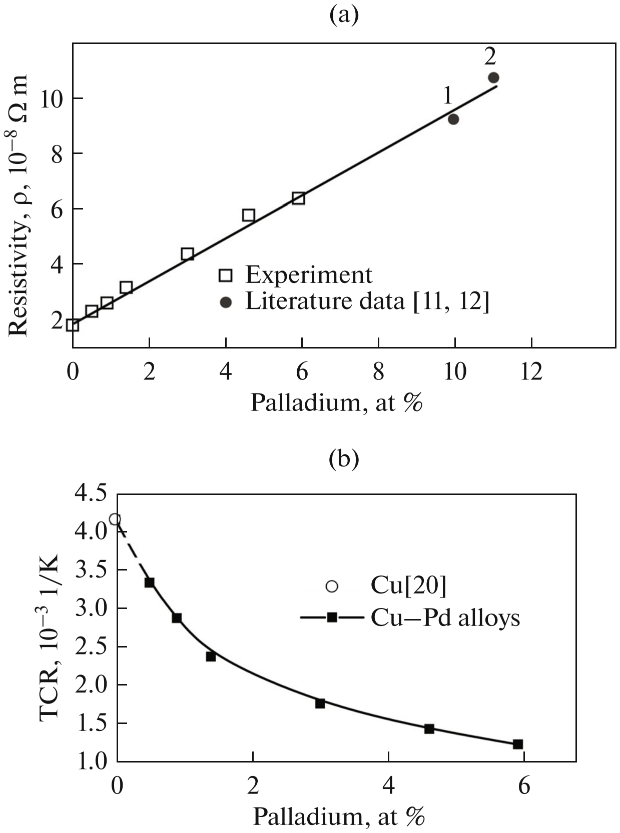

Fig. 5. Concentration dependences of (a) electrical resistivity and (b) temperature coefficient of resistivity of the investigated Cu–Pd alloys at room temperature (obtained based on the data of Fig. 4).

To determine the fine structure, we measured the resistivity in alloys with a high content of palladium, such as Cu–4.6Pd (Fig. 6a) and Cu–5.9Pd (Fig. 6b).

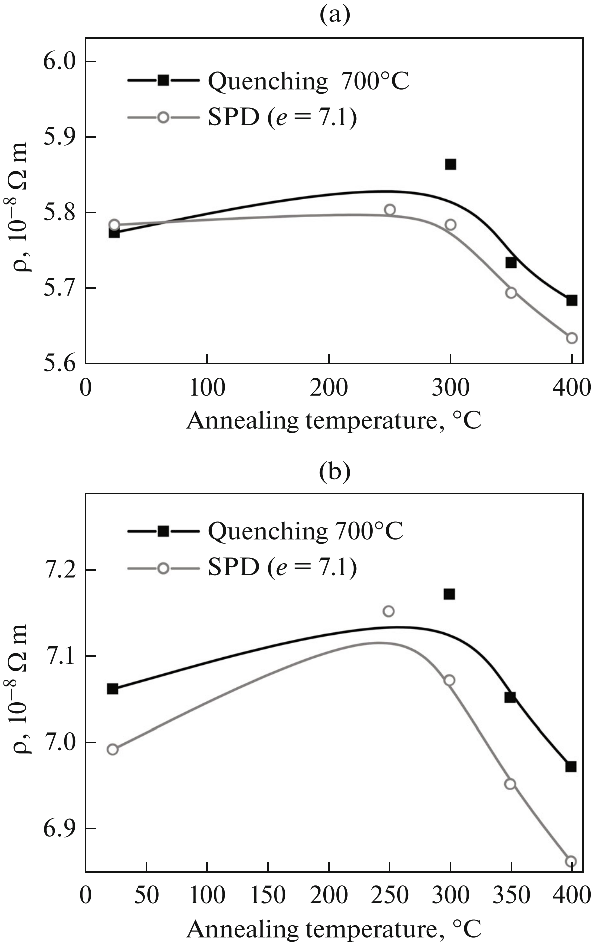

Fig. 6. Changes in the resistivity of (a) Cu–4.6Pd and (b) Cu–5.9Pd alloys after annealing for 168 h. Before annealing, the samples of the alloys were in (■) initial quenched state and (○) preliminarily deformed (to $e \approx 7.1$) state.

The change in the resistivity was studied after annealing the samples for 168 h in the temperature range of 200–400°C (with a step of 50 K). The samples of the alloys were taken in two initial states, i.e., after severe deformation by drawing ($e \approx 7.1$) and after quenching from 700°C.

In the course of experiments, it was revealed that the resistivity of the Cu–4.6Pd alloy (Fig. 6a) almost does not depend on the initial state of the samples (quenched or deformed). The deformed sample of the Cu–5.9Pd alloy has a lower resistivity in comparison with the quenched state (Fig. 6b). This result shows that the evolution of the microstructure in the course of the plastic deformation of the Cu–5.9Pd alloy differs significantly from the processes that are developed upon the deformation of pure copper. Indeed, an increase in the defect state of the structure of copper in the course of severe plastic deformation naturally leads to an increase in the resistivity by approximately 6% in comparison with the annealed state [25]. In Fig. 6b, the inverse dependence is observed.

The changes in the resistivity of the Cu–4.6Pd and Cu–5.9Pd alloys after low-temperature annealing observed in Fig. 6 require additional analysis. For example, it could be expected that the annealing of samples that were initially both in the deformed and quenched states can lead only to a drop in the electrical resistance. Indeed, the quenching from a high temperature produces an excess concentration of vacancies, and a prolonged low-temperature annealing makes it possible to decrease the amount of these defects [19]. The influence of annealing on the resistivity of the strongly deformed samples was considered above based on the example of copper. However, it can easily be seen from Fig. 6a that the low-temperature annealings lead to a certain increase in the resistivity of the quenched Cu–4.6Pd alloy. The expected reduction in the resistivity of the quenched alloy begins after treatments at higher temperatures. The resistivity of the initially deformed sample of this alloy behaves as predicted; at a certain stage of heat treatment, processes of recrystallization are developed that cause a decrease in ρ.

In turn, the annealings at temperatures of 250 and 300°C noticeably increase the resistivity of the samples of the Cu–5.9Pd alloy, regardless of its initial state (Fig. 6b). For a preliminarily quenched alloy, this increase is ~1.5%. An increase in the resistivity of the quenched Cu–4.6Pd and Cu–5.9Pd alloys observed in Fig. 6 can be explained by the formation of clusters of short-range order during the quenching and by an increase in their quantity in the course of low-temperature annealing, as was done earlier [19, 23, 24]. Note that the X-ray diffraction analysis did not reveal any changes in the phase composition of this alloy upon the annealing after SPD (Fig. 2). However, the difference in the behavior of the Cu–5.9Pd alloy compared to other investigated alloys was noted above with consideration of the temperature dependences of the resistivity (see Fig. 4).

The change in the resistivity upon the annealing of the deformed Cu–5.9Pd alloy (Fig. 6b) also requires explanation. Furthermore, it can be seen from Fig. 6 that, even after prolonged holding of the alloys at a temperature of 400°C, significant disagreement is observed between the values of the resistivity of samples that were in different initial states. We also observed similar dependences earlier during the study of Cu–Pd alloys near the equiatomic composition, in which a very slow $A1 \rightarrow B2$ phase transformation occurs [14]. A preliminary plastic deformation significantly accelerates the reconstruction of the structure; therefore, the reduction in the resistivity of the deformed samples caused by atomic ordering occurs more rapidly. Therefore, it can be assumed that the state of equilibrium in the Cu-4.6Pd and Cu-5.9Pd alloys was not achieved, even after annealing for a week at a temperature of 400°C (Fig. 6).

Thus, the results given in Fig. 6 can be satisfactorily explained by the formation of clusters of short-range order in quenched Cu–4.6Pd and Cu–5.9Pd alloys. The formation of small nuclei of the ordered phase can lead to an increase in the resistivity observed upon heating of the initially deformed Cu–5.9Pd alloy (see Fig. 4 and Fig. 6b). Based on the data obtained, it can be assumed that the phase composition of the Cu–Pd alloys investigated (especially with an enhanced content of palladium) is more complicated than follows from the phase diagram.

Mechanical Properties

The concentration dependence of the strength properties of the samples of the investigated Cu–Pd alloys is presented in Fig. 7. As was mentioned above, for comparison we also conducted mechanical tests of pure copper. The initial state of all samples was the state after a preliminary deformation by drawing to 90% ($e \approx 2.3$). It follows from the results obtained that even minimum alloying with palladium leads to a significant increase in the yield stress (~20%) compared with pure copper, the proof yield stress of which was $\sigma_{0.2} = 350$ MPa. With a further increase in the content of palladium, the yield stress of the alloys increases gradually and reaches $\sigma_{0.2} = 505$ MPa in alloy 6 with a Pd content of 5.9 at %. The ultimate strength of all samples also increases. It can be seen from Fig. 7 that, in the range of 0.5–4.6 at % Pd, an almost linear increase is observed in the strength properties. This dependence is in good agreement with the concept of solid-solution strengthening.

It was discovered in the course of experiments that the introduction of palladium substantially increases the recrystallization temperature. The changes in the yield stress ($\sigma_{0.2}$) and in elongation at rupture (δ) after holding of the preliminarily deformed samples of the Cu–0.9Pd and Cu–3.0Pd alloys for 1 h in the range of temperatures of 100–450°C is shown in Figs. 8a and 8b, respectively. The results of the mechanical tests of samples of pure copper are also given here for comparison. As follows from the data obtained, in order to remove the work hardening in the copper samples, annealing at a temperature of 150°C for 1 h is sufficient. The introduction of 0.9 at % Pd increases the recrystallization temperature to 300°C. An increase in the content of palladium in the alloy to 3 at % raises the recrystallization temperature by 50 K more (Fig. 8a). The plastic properties of all alloys after the completion of the processes of recrystallization are comparable as follows: δ ≈ 50% (Fig. 8b).

As is known, the thermal stability of the structure of deformed copper is low: the processes of recrystallization occur even at room temperature [26, 27]. The introduction of palladium leads to a substantial increase in the thermal stability of the structure of the alloys in the course of prolonged low-temperature treatments. Figure 9 demonstrates the dependences of the yield stress of the Cu–0.5Pd and Cu–1.4Pd alloys on the time of holding at a temperature of 250°C. It can be seen well that no substantial changes in the

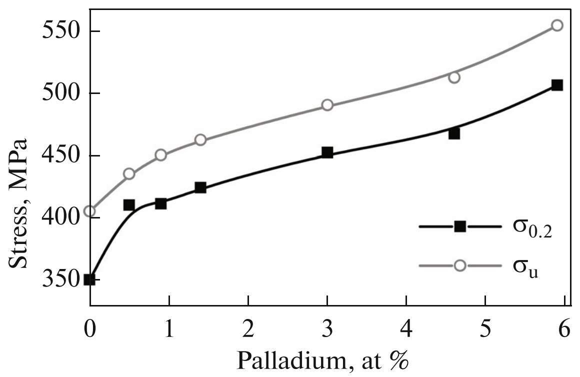

Fig. 7. Concentration dependences of the yield stress ($\sigma_{0.2}$) and ultimate strength (σ_u) of the samples of the investigated alloys (initial state after deformation to $e \approx 2.3$.

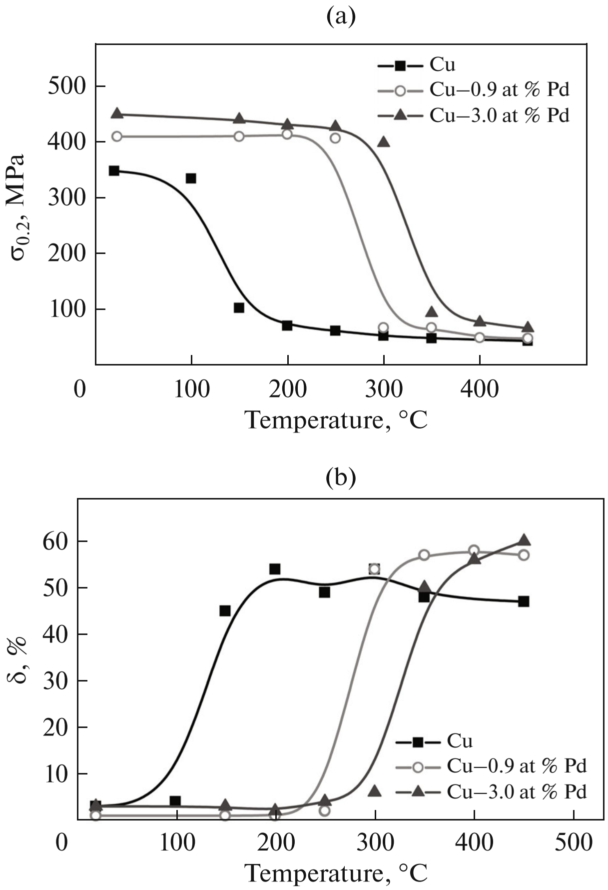

Fig. 8. Changes in (a) yield stress and (b) elongation at rupture of the initially deformed ($e \approx 2.3$) samples of (■) copper and (○) Cu–0.9Pd alloy and (▲) Cu–3.0Pd alloy after annealing in the range of temperatures of 100–450°C for 1 h.

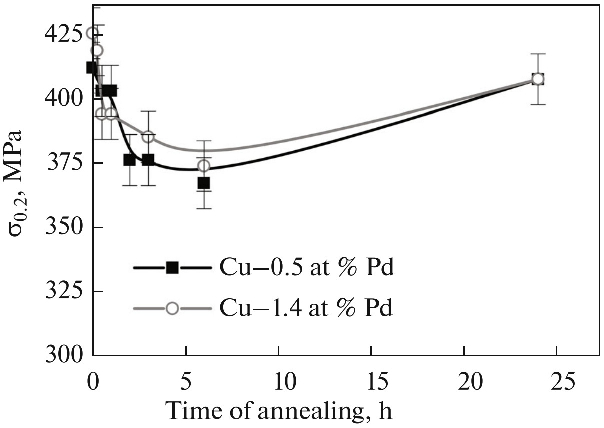

Fig. 9. Variation of the yield stress of the initially deformed ($e \approx 2.3$) samples of the (■) Cu–0.5Pd and (○) Cu–1.4Pd alloys depending on the time of holding at a temperature of 250°C.

strength properties occur, even after holding of the maximum duration (for 24 h). In this case, the plasticity of samples also hardly changes; elongation at rupture is $\delta \approx 3–5\%$. Let us recall that, for the completion of the processes of recrystallization in pure copper, treatment for 1 h at a temperature of 150°C is sufficient (Fig. 8). The totality of the physicomechanical properties obtained in this work upon the study of the low-alloy Cu–Pd alloys is presented in Table 2.

DISCUSSION OF RESULTS

In the course of the above investigation, interesting scientific results were obtained, which are also of specific practical interest. For example, the assumption was made that, in Cu–4.6Pd and Cu–5.9Pd alloys, the formation of clusters of a short-range order or a certain number of nuclei of an ordered phase can occur. This result differs from those predicted by the conventional phase diagram; therefore, it requires additional checking. An analysis of the literature sources showed that, earlier [28], the formation of an atomic short-range order in the quenched alloy Cu–Pd containing 12.6 at % Pd was observed. No publications devoted to the study of the microstructure of alloys with the lower content of palladium have been revealed. Thus, conducting an electron-microscopic examination of the alloys in different structural states is of undoubted scientific interest. We plan to continue our work in this direction.

As was already noted above, an active search for copper-based conductors that have enhanced strength properties and a high electrical conductivity is carried out at present. The results obtained have shown that the strength properties of copper can be somewhat increased by the introduction of palladium. From the viewpoint of using the obtained results in practice, alloys with the minimum content of palladium are of greatest interest (Table 2). Indeed, the addition of palladium in the amount of only a few tenths of percent makes it possible to increase the yield stress by approximately 20% compared with pure copper with a minimum loss of electrical conductivity (76% IACS). It has also been shown in the work that the structure obtained has a higher thermal stability. Note that the corrosion resistance of this alloy is also higher than that of pure copper.

It should here be mentioned that, as a rule, to significantly strengthen copper-based alloys, different methods of SPD are usually employed [4–6]. In our study, we did not have this opportunity; the diameter of our sample after homogenization was 5 mm and the samples for the mechanical tests had a diameter of 1.5 mm. Thus, the true deformation of the samples for mechanical tests in this work was $e \approx 2.3$ or ~90% obtained by drawing. On the one hand, this treatment can easily be realized in practice. At the same time, it is obvious that the strength properties of the investigated alloys can be increased somewhat more using preliminary deformation to higher degrees.

It could be proposed using copper–palladium matrix as the starting material for using other, already approved methods of strengthening copper. One of these methods is precipitation hardening upon alloying with silver. As follows from the results obtained in [29], the use of silver as an addition for alloys with a higher content of palladium leads to significant strengthening due to the precipitation of an Ag–Pd-based phase. Furthermore, the high-strength resistive

Table 2. Physicomechanical properties of copper and Cu–Pd alloys after preliminary deformation

Alloy	Content of Pd, at %	$\sigma_{0.2}$, MPa	σ_u, MPa	ρ, 10^{-8} Ω m	IACS, %
–	0	350	405	1.78	97
1	0.5	400	435	2.28	76
2	0.9	410	450	2.58	67
3	1.4	420	460	3.14	55
4	3.0	450	490	4.35	40
5	4.6	475	510	5.75	30
6	5.9	500	555	6.35	27

Cu–Pd–Ag alloys have found application in practice, e.g., in the domestic automobile industry as springs in the current-collecting position pickups of throttles [30]. The replacement of the high alloy used for these purpose by an alloy with a substantially lower content of palladium and silver with comparable physicomechanical properties is an important practical and scientific problem.

CONCLUSIONS

It has been established by X-ray diffraction methods that, in the investigated Cu–Pd alloys, an fcc solid solution is formed; no presence of any other phases has been revealed. The study of a fine structure using resistivity measurements has shown that in Cu–4.6 at % Pd and Cu–5.9 at % Pd alloys atomic clusters with a short-range order or nuclei of an ordered phase can be formed. The addition of palladium to copper in amounts of a few tenths of percent makes it possible to expand the temperature interval of structural stability and increase the yield stress by ~20% with a minimum loss of electrical conductivity.

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