The development of powder metallurgy Mo–Cu–Ni alloys possessing predetermined thermal expansion characteristics has in recent years focused attention on this ternary system [1–3]. To be able to exercise control over the structure and properties of such alloys by performing synthesis and heat treatment at suitable temperatures, it is necessary to have an accurate knowledge of the phase constitution of the Mo–Cu–Ni system.

The synthesis of such a ternary alloy is effected by sintering a mixture of powders at temperatures lying above the liquidus of the copper–nickel constituent of the material. Under these conditions, molybdenum passes into the resultant liquid copper–nickel phase, while copper and nickel dissolve in the molybdenum. The existence in the system of diffusional mass transport involving dissolution and reprecipitation is clearly confirmed by the fact that the molybdenum phase grains markedly increase in size and at the same time become less numerous in materials characterized by high volume concentrations of the copper–nickel phase (Fig. 1).

Our investigations have shown that the sintering, at temperatures of up to 1200°C and holding times of up to 25 h, of composites with 7–70 wt.% copper–nickel phase contents at nickel concentrations in the copper ranging from 0 to 25% produces no new phases. After heat treatment at temperatures in excess of 1250°C, new phases can be detected by x-ray diffraction and metallographic methods in composites containing more than 9% Ni.

Single-Phase Molybdenum Field. The solubility of nickel in molybdenum in the solid state is negligible, attaining ∼0.5 wt.% at 1200°C [4, 5]. According to [4], molybdenum and copper exhibit no intersolubility in the whole range of concentration. The absence of intersolubility between molybdenum and copper has been noted by many authors. In one investigation, however, the existence of a molybdenum base solid solution containing up to 1.5 wt.% copper was established as a result of electrical conductivity and lattice constant measurements on the phases of Mo–Cu specimens sintered at 1600–1700°C, annealed, and quenched from 950°C [6]. The existence of solubility in the system Mo–Cu was confirmed by our own studies [7], which demonstrated that the dissolution of copper in molybdenum reduces the lattice constant of the latter from 3.1412 (for the starting pure powder) to 3.1408 ± 0.0001 kX.

For a study of mutual solubility in the system under consideration, a 90 Mo–8 Cu–2 wt.% Ni alloy sintered at 1250°C was chosen. Determinations were made, using the technique described in [7], of changes in the lattice constant of the molybdenum phase of the alloy on the basis of data yielded by x-ray diffraction analyses of alloy fillings subjected to annealing and quenching from 950°C. The molybdenum lattice constant was found to decrease to 3.1403 ± 0.0001 kX.

The reaction of molybdenum particles with copper and nickel causes their microhardness to rise from 240 kg/mm² for pure sintered molybdenum powder to 300 kg/mm² for a binary sintered Mo–Cu composite and 380–400 kg/mm² for a ternary alloy of the composition given above. Nickel content variations lead to corresponding changes in the microhardness of the molybdenum phase, as illustrated in Fig. 2a (curve 1).
The existence of a relationship between the microhardness of the Mo phase and the nickel and copper contents of these alloys is confirmed by the fact that the microhardness of sintered specimens infiltrated for 1 h in a copper-nickel melt containing 5% Ni (relative to copper) is 10-30% less than that of sintered specimens. By filling the pores, such infiltration raises the nickel content of the alloy, although the concentration of nickel falls relative to copper. Figure 2b shows the correlation between the composition of the copper-nickel phase in ternary Mo-Cu-Ni composites and the microhardness of their molybdenum phase.

Boundary of Solubility of Molybdenum in Ni-Cu Solid Solutions. According to [4, 5], the solubility of molybdenum in a solid solution of nickel is 36 wt.% at 1200°C. Molybdenum is insoluble in solid and liquid copper [4]. The work described below was undertaken with the aim of determining the boundary of solubility of molybdenum in Ni-Cu solid solutions.

Two series of alloys were prepared for investigation. Alloys of Series A were obtained by pressing and sintering powders and remelting the resultant specimens in zirconia crucibles in a resistance furnace provided with a hydrogen atmosphere. As starting materials, molybdenum (99.92 Mo, 0.031 Ni, 0.006 Al₂O₃, 0.003 Fe₂O₃, and 0.03 wt.% SiO₂), copper (99.95 Cu, 0.02 Zn, 0.013 Ni, 0.0007 Fe, 0.01 Pb, and 0.005 wt.% Sn), and nickel (99.66 Ni, 0.002 Fe, 0.001 Cu, 0.001 Mn, 0.001 Pb, and 0.001 wt.% Sn) powders were used. Alloys of Series B were produced by the technique of induction levitation melting in a hydrogen atmosphere, using quartz glass shells, from weighed samples of copper (vacuum melted), nickel, and molybdenum foil (vacuum melted). The total impurity contents of these components did not exceed 0.01, 0.1, and 0.07%, respectively. Data upon the charge compositions of test alloys and the heat treatments applied to specimens are presented in Table 1. The Series A alloys were annealed in alumina crucibles in a hydrogen atmosphere, and the Series B alloys in sealed quartz capsules in a hydrogen furnace.

For analysis, slugs (weighing 10 g each) were cut in two along their vertical axes. As-cast and heat treated specimens were examined metallographically and subjected to X-ray diffraction and spectral analyses, and their hardness was measured. To bring out their microstructures, prepared sections were etched with a mixture of acetic and nitric acids (1:1). X-ray structural analyses of microsections were performed in URS-50I and URS-55A diffractometers, using copper Kα radiation. The amounts of components (Ni and Mo) in the homogeneous parts of alloys were determined quantitatively by an emission microspectral analysis technique similar to that described in [8]. Nickel and molybdenum reference specimens were prepared from the starting materials employed in the investigation.

Second phase precipitates in both the alloys obtained from powders (Series A) and those produced by melting bulk metals (Series B) were found to be concentrated in the slug centers, where they formed, at greater volumes of the phase, a dense nucleus (Fig. 3). This feature of structure formation made it possible...