

METALLIC SURFACES AND FILMS

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Influence of Copper on $A1$ to $L1_0$ Phase Transformation in Nanoscale $\text{Fe}_{50}\text{Pt}_{50}$ Films

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By the methods of materials science, the effect of intermediate Cu layer with low surface energy ($\cong 1.83 \text{ J/m}^2$) (top, intermediate, and underlayer) in $[\text{Fe}_{50}\text{Pt}_{50}(15 \text{ nm})/\text{intermediate Cu}(7.5 \text{ nm}) \text{ layer}/\text{Fe}_{50}\text{Pt}_{50}(15 \text{ nm})]_n$ (where $n = 1, 2$), top Cu(7.5 nm) layer/ $\text{Fe}_{50}\text{Pt}_{50}(15 \text{ nm})$ and $\text{Fe}_{50}\text{Pt}_{50}(15 \text{ nm})/\text{underlayer Cu}(7.5 \text{ nm})$ film compositions on $\text{SiO}_2(100 \text{ nm})/\text{Si}(001)$ substrates on diffusion phase-formation processes and $L1_0$ phase formation, its structure, and magnetic properties at annealing in vacuum is studied. The film compositions are prepared by magnetron sputtering on thermally oxidized SiO_2 layer by thickness of 100 nm on monocrystalline Si(001) substrate. Subsequent heat treatment is carried out at high vacuum of $1.3 \cdot 10^{-3} \text{ Pa}$ in the 300–900°C temperature range during 30 s at each temperature. As determined, the chemically disordered $A1(\text{FePt})$ phase is formed in all as-deposited films. The formation of chemically ordered $L1_0(\text{FePt})$ phase in $[\text{Fe}_{50}\text{Pt}_{50}(15 \text{ nm})/\text{Cu}(7.5 \text{ nm}) \text{ intermediate layer}/\text{Fe}_{50}\text{Pt}_{50}(15 \text{ nm})]_n$ films (where $n = 1, 2$) with intermediate layers takes place during annealing at 700°C and is accompanied by sharp coercivity increase, which also rises after subsequent high-temperature annealing. In the films with top copper layer, the temperature of $L1_0(\text{FePt})$ phase formation rises up to 900°C. In the films with copper underlayer, the formation of $L1_0(\text{FePt})$ phase is not detected by X-ray analysis, but small coercivity increasing after annealing within the temperature range of 800–900°C can testify that ordering processes proceed.

Методами фізичного матеріалознавства вивчено вплив додаткового шару Cu з низькою поверхневою енергією ($\cong 1,83 \text{ Дж/м}^2$) (верхнього, проміжного і підшару) в плівкових композиціях $[\text{Fe}_{50}\text{Pt}_{50}(15 \text{ нм})/\text{проміжний шар Cu}(7,5 \text{ нм})/\text{Fe}_{50}\text{Pt}_{50}(15 \text{ нм})]_n$ (де $n = 1, 2$), верхній шар Cu(7,5 нм)/ $\text{Fe}_{50}\text{Pt}_{50}(15 \text{ нм})$ і $\text{Fe}_{50}\text{Pt}_{50}(15 \text{ нм})/\text{підшар Cu}(7,5 \text{ нм})$ на підложжях $\text{SiO}_2(100 \text{ нм})/\text{Si}(001)$ на процеси дифузійного фазоутворення, формування фази $L1_0$ та її структурні й магнетні властивості при відпалах у вакуумі. Плівкові

композиції одержано методом магнетронного осадження на термічно окиснене (шар SiO_2 товщиною 100 нм) підложжя монокристалічного $\text{Si}(001)$. Наступне термічне оброблення тривалістю у 30 секунд виконувалося у високому вакуумі $1,3 \cdot 10^{-3}$ Па в температурному інтервалі 300–900°C. Встановлено, що в усіх плівках після осадження формується хемічно неупорядкована фаза $A1(\text{FePt})$. Формування хемічно впорядкованої фази $L1_0(\text{FePt})$ у плівках з проміжними шарами міді $[\text{Fe}_{50}\text{Pt}_{50}(15 \text{ нм})/\text{проміжний шар Cu}(7,5 \text{ нм})/\text{Fe}_{50}\text{Pt}_{50}(15 \text{ нм})]_n$ (де $n = 1, 2$) відбувається в ході відпалу при температурі 700°C і супроводжується різким збільшенням коерцитивної сили, яка зростає також і після наступних високотемпературних відпалів. У плівках з верхнім шаром міді температура формування фази $L1_0(\text{FePt})$ підвищується до 900°C. У плівці з підшаром міді утворення фази $L1_0(\text{FePt})$ рентгенографічно не встановлено, але невелике збільшення коерцитивної сили після відпалів в інтервалі температур 800–900°C може свідчити про перебіг процесів упорядкування.

Методами фізического матеріалознавства вивчено вплив додаткового шару Cu з низькою поверхневою енергією ($\cong 1,83 \text{ Дж/м}^2$) (верхнього, проміжного і підшару) в пліночних композиціях $[\text{Fe}_{50}\text{Pt}_{50}(15 \text{ нм})/\text{проміжний шар Cu}(7,5 \text{ нм})/\text{Fe}_{50}\text{Pt}_{50}(15 \text{ нм})]_n$ (де $n = 1, 2$), верхній шар $\text{Cu}(7,5 \text{ нм})/\text{Fe}_{50}\text{Pt}_{50}(15 \text{ нм})$ і $\text{Fe}_{50}\text{Pt}_{50}(15 \text{ нм})/\text{підшар Cu}(7,5 \text{ нм})$ на підложках $\text{SiO}_2(100 \text{ нм})/\text{Si}(001)$ на процеси дифузійного фазоутворення, формування фази $L1_0$ і її структурні і магнітні властивості при отжигу в вакуумі. Пліночні композиції отримані методом магнетронного осадження на термічно окиснену (шар SiO_2 товщиною 100 нм) підложку монокристалічного $\text{Si}(001)$. Наступна термічна обробка тривалістю 30 секунд виконувалася в високому вакуумі $1,3 \cdot 10^{-3}$ Па в температурному інтервалі 300–900°C. Встановлено, що в усіх плівках після осадження формується хемічно неупорядкована фаза $A1(\text{FePt})$. Формування хемічно упорядкованої фази $L1_0(\text{FePt})$ в плівках з проміжними шарами міді $[\text{Fe}_{50}\text{Pt}_{50}(15 \text{ нм})/\text{проміжний шар Cu}(7,5 \text{ нм})/\text{Fe}_{50}\text{Pt}_{50}(15 \text{ нм})]_n$ (де $n = 1, 2$) відбувається в процесі отжигу при температурі 700°C і супроводжується різким збільшенням коерцитивної сили, яка зростає також і після наступних високотемпературних відпалів. У плівках з верхнім шаром міді температура формування фази $L1_0(\text{FePt})$ підвищується до 900°C. У плівці з підшаром міді утворення фази $L1_0(\text{FePt})$ рентгенографічно не встановлено, але невелике збільшення коерцитивної сили після отжигу в інтервалі температур 800–900°C може свідчити про перебіг процесів упорядкування.

Key words: chemically ordered phase $L1_0(\text{FePt})$, thin films, annealing, coercive force.

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1. INTRODUCTION

Using of nanosize FePt films with chemically ordered high-coercivity

phase L1₀(FePt) makes it possible to increase the density of magnetic recording and information storage up to $\cong 1$ Tb/cm². High energy of magnetic-crystalline anisotropy of the L1₀(FePt) phase $K_u = 7 \cdot 10^6$ J/m³, which prevents transition to superparamagnetic state with the decrease of grain volume, promotes it. L1₀(FePt) phase is formed from the chemically disordered magnetically soft phase A1(FePt) at temperatures higher than 400°C. To enhance a technological effectiveness at usage of these films, the temperature of L1₀(FePt) phase formation should be decreased. One of the way to accelerate the ordering process is the using of the energy of the boundaries, which are additionally formed in the film composition as a result of introduction of additional layer of the third element with low surface energy (Ag, Cu, Au) [1–13]. It is supposed that the low surface energy in layered film compositions can be used as extra driving force, which promotes diffusion rearrangement of Fe and Pt atoms and formation of the chemically ordered L1₀(FePt) phase with necessary magnetically hard properties.

The purpose of this work is investigation of the impact of the additional layer of Cu with its different location in the film composition (upper, intermediate, and sublayer) on the processes of diffusional phase formation and transition of A1(FePt) phase into L1₀(FePt) phase, its structural and magnetic properties in nanosize film compositions [Fe₅₀Pt₅₀(15 nm)/intermediate layer Cu/Fe₅₀Pt₅₀(15 nm)]_n/SiO₂(100 nm)/Si (001) (where $n = 1, 2$), Fe₅₀Pt₅₀(15 nm)/underlayer Cu(7.5 nm)/SiO₂(100 nm)/Si(001) and upper layer Cu(7.5 nm)/Fe₅₀Pt₅₀(15 nm)/SiO₂(100 nm)/Si(001) at annealing in vacuum.

2. EXPERIMENTAL TECHNIQUE

Nanosize film compositions (NFC) [Fe₅₀Pt₅₀(15 nm)/intermediate layer Cu/Fe₅₀Pt₅₀(15 nm)]_n/SiO₂(100 nm)/Si (001) (where $n = 1, 2$), Fe₅₀Pt₅₀(15 nm)/underlayer Cu(7.5 nm)/SiO₂(100 nm)/Si(001) and upper layer Cu(7.5 nm)/Fe₅₀Pt₅₀(15 nm)/SiO₂(100 nm)/Si(001) are produced by the method of layer-by-layer deposition of the layers of Fe₅₀Pt₅₀ (99.95%) alloy with the thickness of 15 nm and layers of Cu (99.9%) with the thickness of 7.5 nm on the substrate of the thermally oxidized layer of SiO₂ with the thickness of 100 nm on Si single crystal of (001) orientation, which is at the room temperature. The annealing is performed in the high vacuum at the pressure of $\cong 1.3 \cdot 10^{-3}$ Pa within the temperature range of 300–900°C with the holding of 30 sec at each temperature at the heating rate of 5°C/s. The cooling rate is 0.25°C/s. The thickness of the deposited layer is measured using a quartz crystal resonator and by the method of X-ray reflectometry. The error of the thickness estimate is ± 1 nm.

The investigation of NFC after deposition and annealing and determination of the degree of their chemical ordering is performed by the

X-ray diffraction method using ULTIMA IV Rigaku diffractometer in $\text{CuK}\alpha$ -radiation. The degree of chemical ordering of the $L1_0(\text{FePt})$ phase is estimated by the ratio of the diffraction reflexes $I(001)/I(002)$ [14, 15]. The degree of orientation of easy magnetic axis [001] with regard to normal to a surface is determined by the ratio $I(001)/I(111)$. The change of the chemical composition across the film thickness because of diffusion processes is investigated by the method of Rutherford back scattering (RBS).

Morphology of film surface is investigated by an atomic force microscopy (AFM). Magnetic properties of the films are estimated with the help of the Kerr magneto-optical effect and magnetic force microscopy. Resistometric measurements are performed by four-point probe technique at room temperature.

3. RESULTS AND DISCUSSION

Diffractograms of NFC with different locations of the copper layer af-

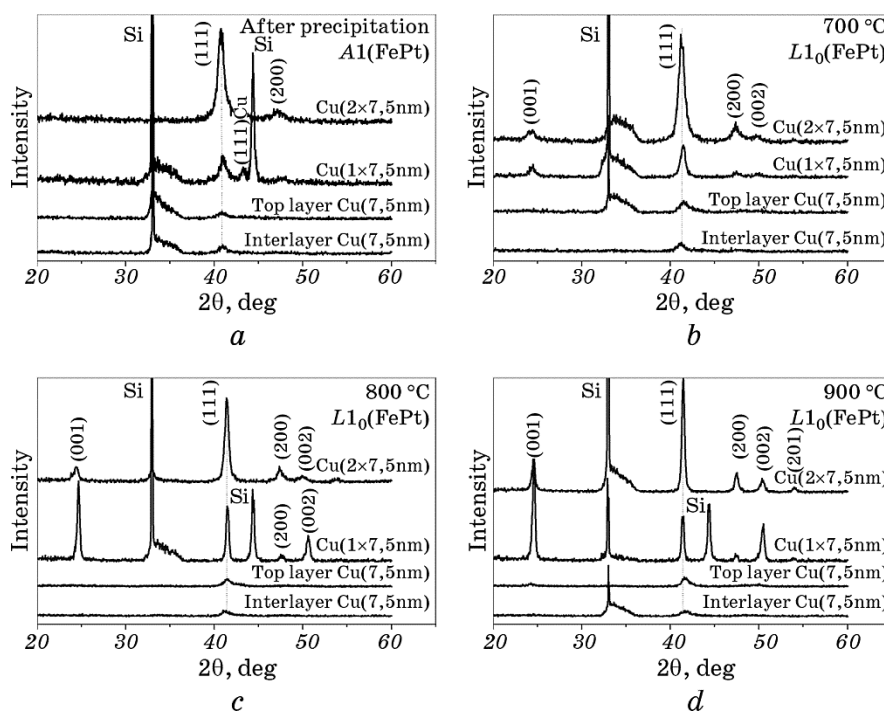


Fig. 1. XRD patterns (for $\text{CuK}\alpha$ -radiation) of $[\text{Fe}_{50}\text{Pt}_{50}(15 \text{ nm})/\text{Cu}(7,5 \text{ nm})/\text{Fe}_{50}\text{Pt}_{50}(15 \text{ nm}) \text{ intermediate layer}]_n$ (where $n = 1, 2$), $\text{Fe}_{50}\text{Pt}_{50}(15 \text{ nm})/\text{Cu}(7,5 \text{ nm})$ underlayer and $\text{Cu}(7,5 \text{ nm})$ top layer/ $\text{Fe}_{50}\text{Pt}_{50}(15 \text{ nm})$ as-deposited films (a) and after annealing at temperatures of 700°C (b), 800°C (c), 900°C (d).

ter deposition and annealing in the temperature range 700–900°C are presented in Fig. 1.

After deposition, structural reflexes (111) from the chemically disordered phase A1(FePt) are fixed in all films, from copper and from the substrate (see Fig. 1, *a*). The value of the coercive force of all films comprises about 50 oersted that testifies their soft-magnetic properties, which remain unchanged until the annealing temperature of 600°C (see Fig. 2).

According to the results of the X-ray diffraction phase analysis, annealing of the investigated films within the temperature range of 300–600°C are not accompanied by the significant change of their structural and phase composition and magnetic properties. However, as noted in paper [16], results of investigation of the films with the intermediate copper layer [Fe₅₀Pt₅₀(15 nm)/intermediate Cu(7.5 nm) layer/Fe₅₀Pt₅₀ (15 nm)] by the RBS method show that the thermally activated diffusion processes between the layers of Cu and FePt take place at the annealing at the temperature of 300°C.

Further annealing in vacuum of the films with the intermediate copper layer [Fe₅₀Pt₅₀(15 nm)/intermediate Cu(7.5 nm) layer/Fe₅₀Pt₅₀ (15 nm)]_{*n*}, where *n* = 1, 2, at the temperature of 700°C is accompanied by the transition of the A1(FePt) phase into the chemically ordered L1₀(FePt) phase (see Fig. 1, *b*). The appearance of the superstructural reflex (001) and splitting of the structural reflex (200) into reflexes (200) and (002) testifies to that. Low intensity of the (001) reflex testifies to the fact that the specified annealing temperature is not suffi-

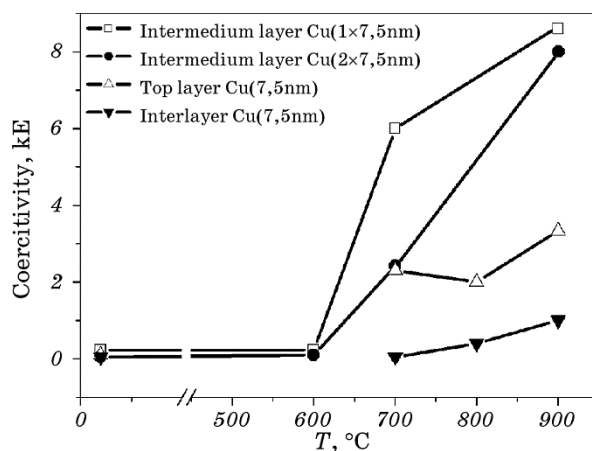


Fig. 2. Dependences of coercivity on annealing temperature for [Fe₅₀Pt₅₀(15 nm)/Cu(7,5 nm)/Fe₅₀Pt₅₀(15 nm) intermediate layer]_{*n*} (where *n* = 1, 2), Fe₅₀Pt₅₀(15 nm)/Cu(7,5 nm) underlayer, and Cu(7,5 nm) top layer/Fe₅₀Pt₅₀(15 nm) films.

cient for the appearance of the complete ordering in the film. The fact that the increase of the annealing temperature is accompanied by the noticeable intensity of (010 and (002) reflexes, serves to show it (see Fig. 1, *c, d*). In this case, their magnetic properties enhance, in particular, the coercive force increases to 8.6 kilooersted ($n = 1$) and 8 kilooersted ($n = 2$) (see Fig. 2).

In films with the upper layer and sublayer of copper, annealing up to and including the temperature of 800°C does not result in significant change of structural and phase state. However, in the film with the upper layer of copper the shift of the structural reflex (111) towards bigger angles is observed, which is a consequence of appearance of tetragonality of the crystal lattice, caused by the start of the phase transformation $A1(\text{FePt}) \rightarrow L1_0(\text{FePt})$. Appearance of the low intensity superstructural reflex (001) of the $L1_0(\text{FePt})$ phase after annealing at the temperature of 900°C, which is by 200°C higher than that for the films with the intermediate copper layer, gives evidence of the lower intensity of the ordering process (see Fig. 1, *c*). In the film with the sublayer of copper, the shift of the structural reflex (111) towards bigger angles is observed only after annealing at the temperature of 900°C, and the superstructural reflex (001) does not appear at all. The presence of the copper sublayer largely decelerates the processes of rearrangement of iron and platinum atoms and shifts the formation of the chemically ordered phase to the higher annealing temperatures. The increase of the coercive force of the film with the upper layer of copper to 2.3 kilooersted only after the annealing at 700°C and of the film with the sublayer of copper to 1 kilooersted after annealing at 900°C also points at the insignificant processes of the ordering in the films under investigation (see Fig. 2).

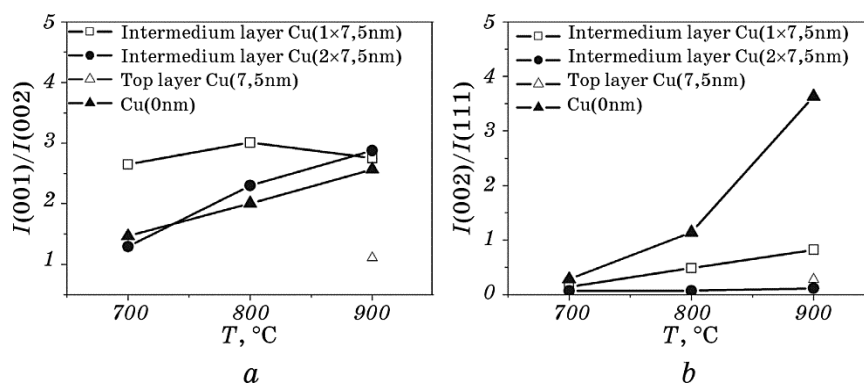


Fig. 3. Dependences of $I(001)/I(002)$ (a) and $I(001)/I(111)$ (b) structural reflexes intensity ratio of $[\text{Fe}_{50}\text{Pt}_{50}(15 \text{ nm})/\text{Cu}(7,5 \text{ nm})/\text{Fe}_{50}\text{Pt}_{50}(15 \text{ nm})$ intermediate layer] $_n$ (where $n = 1, 2$), $\text{Fe}_{50}\text{Pt}_{50}(15 \text{ nm})/\text{Cu}(7,5 \text{ nm})$ underlayer, and $\text{Cu}(7,5 \text{ nm})$ top layer/ $\text{Fe}_{50}\text{Pt}_{50}(15 \text{ nm})$ films.

The ratio of intensities of the structural reflexes (001) and (002) for the films with the intermediate copper layer [Fe₅₀Pt₅₀(15 nm)/intermediate Cu(7.5 nm) layer/Fe₅₀Pt₅₀(15 nm)]_{*n*}, where *n* = 1, 2 after the annealing at the temperature of 700°C reaches values of 2.8 and 1.3, respectively. It gives evidence of the establishment of the long-range order of the L1₀(FePt) phase (see Fig. 3, *a*).

It should be noted that the ratio of intensities of the structural reflexes (001) and (002) for the film with one intermediate copper layer insignificantly changes with the annealing temperature growth. One can suppose that the formation of the chemically ordered phase L1₀(FePt) completed during the annealing at the temperature of 700°C. In the multilayer film with two intermediate layers of copper the degree of ordering increases with temperature (see Fig. 3, *a*). Decrease of the degree of tetragonality *c/a* also points at the presence of the processes of ordering in films with one and two intermediate layers of copper at the annealing temperature increase to 900°C (see Fig. 4, *a*). In this case, as follows from Fig. 4, *a*, more evident decrease of the degree of tetragonality *c/a*, as compared with the film of Fe₅₀Pt₅₀ alloy, points at the fact that introduction of additional layers of copper accelerates the processes of ordering. Insignificant increase of the ratio of intensities of *I*(002)/*I*(111) structural reflexes at annealing temperature increase from 700 to 900°C bears evidence of the turn of the *c* axis in the direction perpendicular to substrate plane of a small number of grains (see Fig. 3, *b*).

The dependences of the crystal lattice parameters *a* and *c* of the FePt phase for the films with one and two intermediate layers of copper are presented in Fig. 4, *b*. Data for the film of Fe₅₀Pt₅₀ alloy are presented for comparison. As one can see from Fig. 4, *b*, after annealing at the temperature higher than 700°C the lattice parameter *c* of L1₀(FePt) phase for the film with intermediate copper layers is lower than that of

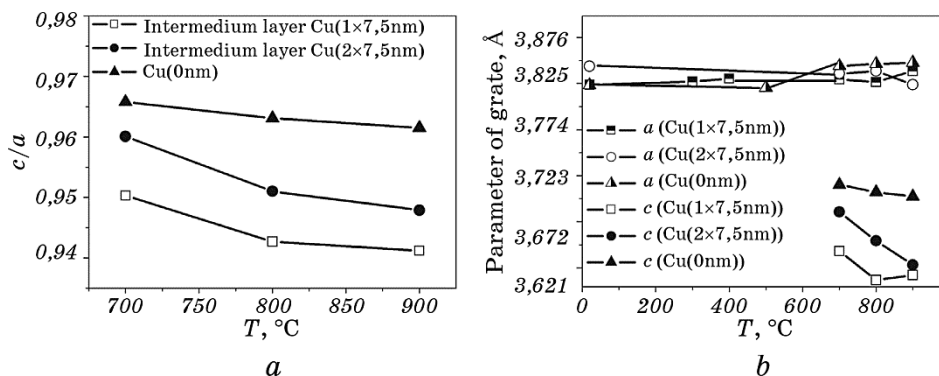


Fig. 4. Dependences of *c/a* ratio (*a*) and *a* and *c* lattice parameters (*b*) of FePt phase on annealing temperature.

the film of the alloy. Decrease of the value of lattice parameter c of the FePt phase for the film alloyed with copper in comparison with that of the pure film points at the formation of the thermally stable ternary alloy Fe(Cu)Pt and the absence of Cu segregating along the grain boundaries of the $L1_0$ (FePt) phase [17]. In addition, the shift of the diffraction peak FePt (001) toward the bigger angles is also serves as evidence of the introduction of copper atoms into the face centred tetragonal lattice of the $L1_0$ (FePt) phase with substitution of Fe or Cu atoms [18]. The diffraction peak (001) of the $L1_0$ (FePt) phase after the annealing at 700°C for the multilayer films $[\text{Fe}_{50}\text{Pt}_{50}(15 \text{ nm})/\text{intermediate Cu}(7.5 \text{ nm}) \text{ layer}/\text{Fe}_{50}\text{Pt}_{50}(15 \text{ nm})]_n$ (where $n = 1, 2$) shifts by 0.539° ($n = 1$) and by 0.579° ($n = 2$), respectively. After the annealing at the temperature of 700°C of the multi-layered films $[\text{Fe}_{50}\text{Pt}_{50}(15 \text{ nm})/\text{intermediate Cu}(7.5 \text{ nm}) \text{ layer}/\text{Fe}_{50}\text{Pt}_{50}(15 \text{ nm})]_n$ during formation of the ternary compound Fe(Cu)Pt the highest copper concentration ($\cong 5\%$) is observed after high temperature annealing in compositions with sublayer and upper copper layer. The absence of all reflexes of the $L1_0$ (FePt) phase can be explained by that the parameter a does not practically change.

The change with annealing temperature of the electrical resistance of the films under investigation and its comparison with the film of the pure $\text{Fe}_{50}\text{Pt}_{50}$ alloy is presented in Fig. 5. As one can see, for all films with the additional copper layer, regardless of its location within the film composition, the increase of the electrical resistance after annealing at the temperature higher than 600°C is observed. It can be con-

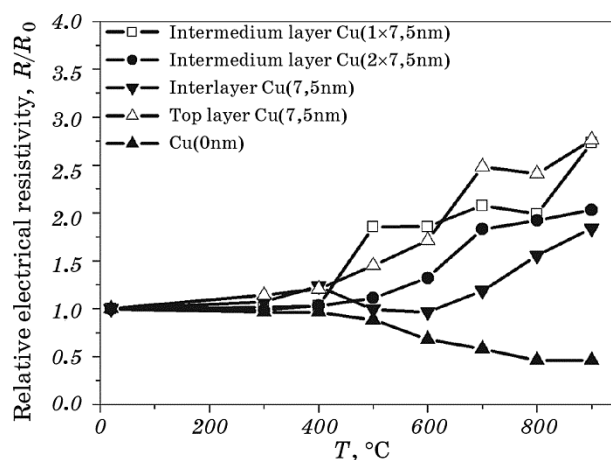


Fig. 5. Dependences of relative electric resistance of $[\text{Fe}_{50}\text{Pt}_{50}(15 \text{ nm})/\text{Cu}(7,5 \text{ nm}) \text{ intermediate layer}/\text{Fe}_{50}\text{Pt}_{50}(15 \text{ nm})]_n$ (where $n = 1, 2$), $\text{Fe}_{50}\text{Pt}_{50}(15 \text{ nm})/\text{Cu}(7,5 \text{ nm})$ underlayer, and $\text{Cu}(7,5 \text{ nm})$ top layer/ $\text{Fe}_{50}\text{Pt}_{50}(15 \text{ nm})$ films on annealing temperature.

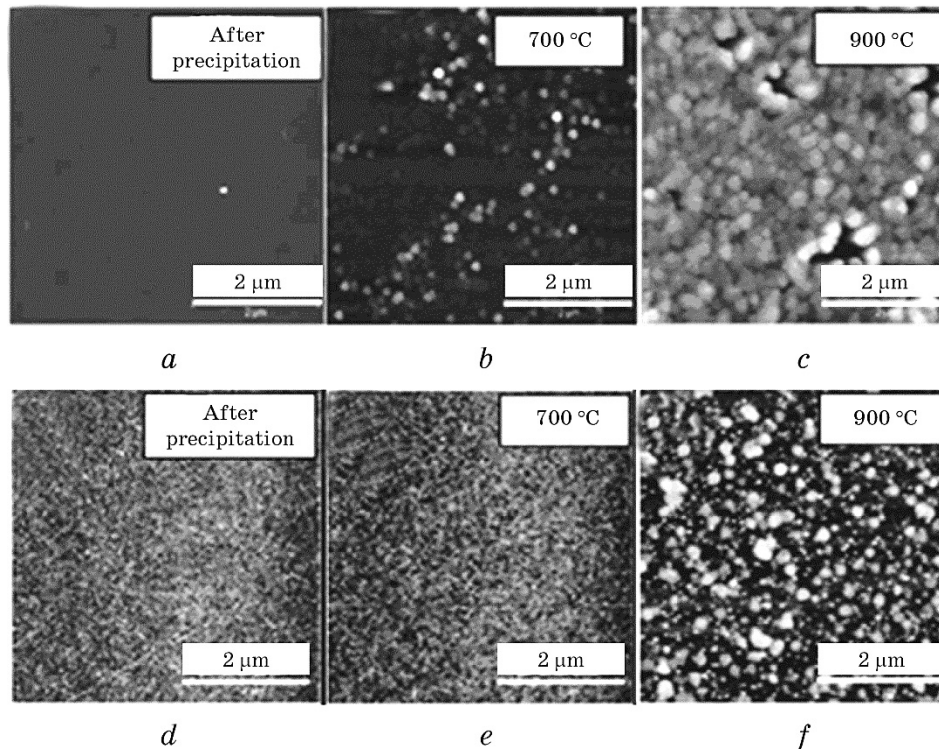


Fig. 6. AFM images of Fe₅₀Pt₅₀(15 nm)/Cu(7.5 nm) intermediate layer/Fe₅₀Pt₅₀(15 nm) (*a–c*) and Fe₅₀Pt₅₀(15 nm)/Cu(7.5 nm) underlayer films (*d–f*) surface morphology after deposition (*a, d*) and thermal treatment at temperature of 700°C (*b, e*) and 900°C (*c, f*).

nected with the diffusion of copper in the grain boundaries and in the lattice of FePt [12] and formation of the ternary compound Fe(Cu)Pt [17].

The morphology of the surface of all investigated films changes noticeably with the annealing temperature increase from 300 to 900°C. Photos of the morphology of the surface of the films Fe₅₀Pt₅₀(15 nm)/intermediate Cu(7.5 nm) layer/Fe₅₀Pt₅₀(15 nm) and Fe₅₀Pt₅₀(15 nm)/intermediate Cu(7.5 nm) layer after deposition and annealing in vacuum are presented in Fig. 6.

As seen from Fig. 6, *a* and Fig 7, *a*, the surface of the deposited films is practically smooth. The annealing temperature increase is accompanied by the growth of surface roughness of the films, in particular, with the intermediate layer of copper (see Fig. 7, *a*), in which noticeable growth of grains of the L1₀(FePt) phase during annealing at the temperatures higher than 600°C is observed (see Fig. 7, *b*). It is connected with the formation of the L1₀(FePt) ordered phase and also with

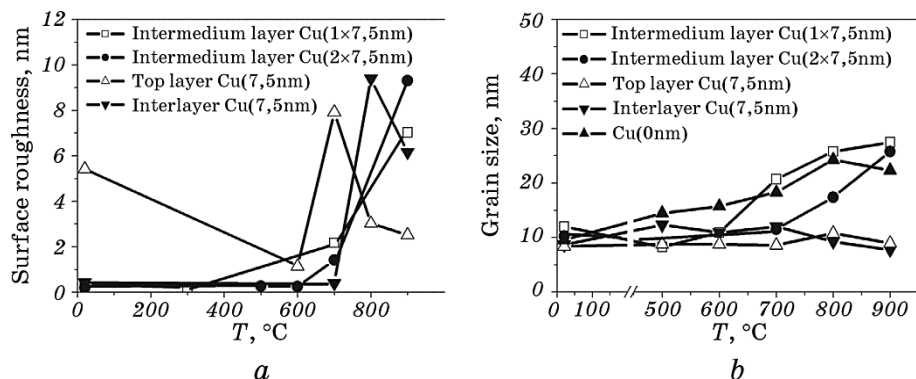


Fig. 7. Dependences of surface roughness (*a*) and grain size on annealing temperature for FePt phase (*b*) in $[\text{Fe}_{50}\text{Pt}_{50}(15\text{ nm})/\text{Cu}(7,5\text{ nm})]$ intermediate layer/ $\text{Fe}_{50}\text{Pt}_{50}(15\text{ nm})$ (where $n = 1, 2$), $\text{Fe}_{50}\text{Pt}_{50}(15\text{ nm})/\text{Cu}(7,5\text{ nm})$ underlayer, and $\text{Cu}(7,5\text{ nm})$ top layer/ $\text{Fe}_{50}\text{Pt}_{50}(15\text{ nm})$ films on $\text{SiO}_2/\text{Si}(001)$ substrate.

the absence of copper segregations on the grain boundaries, which inhibit the grains growth. In the films with the upper and lower Cu layer, in which the formation of the ordered phase is not observed, and the Fe(Cu)Pt ternary compound is formed, the grain size changes insignificantly. One can suppose that copper, which is located on the grain boundaries, suppresses the grain growth.

Magnetic force microscope (MFM) images of the films with the intermediate copper layer $[\text{Fe}_{50}\text{Pt}_{50}(15\text{ nm})/\text{intermediate Cu}(7.5\text{ nm})]$ (where $n = 1, 2$) demonstrate labyrinthine domain structure after annealing at the temperatures of 700°C and 900°C (see

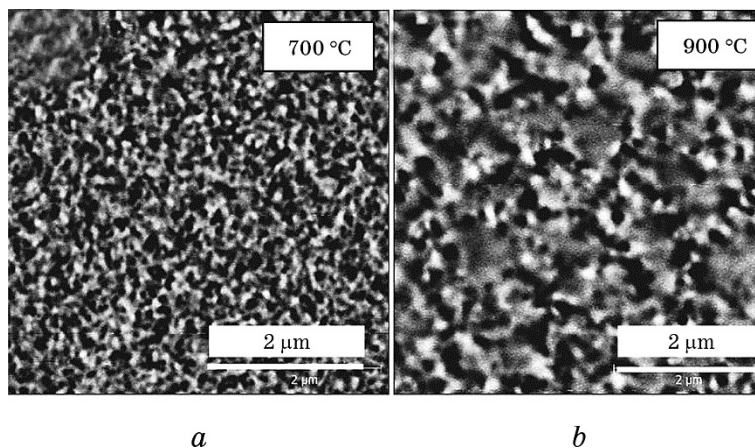


Fig. 8. MFM images of $[\text{Fe}_{50}\text{Pt}_{50}(15\text{ nm})/\text{Cu}(7.5\text{ nm})]$ film after annealing at temperatures of 700°C (*a*), 900°C (*b*).

Fig. 8). MFM images of the multilayer film [Fe₅₀Pt₅₀(15 nm)/intermediate Cu(7.5 nm) layer/Fe₅₀Pt₅₀(15 nm)]₂ are presented in Fig. 8 as an example. It bears witness of the magnetically hard properties of the films under investigation. It is seen that the annealing temperature growth is accompanied by the growth of domains.

The films with domain structure demonstrate maximal values of the coercive force after annealing at the temperature of 900°C (see Fig. 2).

Consequently, the decrease of the ordering temperature in the films with one and two intermediate copper layer by 200°C (from 900°C to 700°C) in comparison with the films with the upper layer and sublayer of copper can be connected with the higher influence of the surface energy on the Fe₅₀Pt₅₀/Cu interface as the extra driving force of the processes of the diffusion phase formation, on one hand. On the other hand, one can suppose that after the high temperature annealing in the films with the upper and lower additional copper layer the elevated copper concentration (\cong 50 at.% Cu) as compared with films with intermediate copper layers (\cong 25 at.% Cu) does not contribute to the decrease of the ordering temperature, however, causes the formation of the FeCuPt ternary compound. This conclusion is in good agreement with the results of work [17], where it is stated that the copper concentration of 15 at.% is the most efficient for the decrease of the ordering temperature.

4. CONCLUSIONS

It is established that the introduction of the additional layers of copper with low surface energy into the film compositions differently influences over the start temperature of the ordering process.

Formation of two or more interfaces in the film composition [Fe₅₀Pt₅₀(15 nm)/intermediate Cu(7.5 nm) layer/Fe₅₀Pt₅₀(15 nm)]_n, where $n = 1, 2$, contributes to the reduction of the ordering temperature by 200°C (from 900°C to 700°C) as compared with the film with the upper copper layer. Formation of the chemically ordered L1₀(FePt) phase is accompanied by the sharp increase of the coercive force of the films after annealing within the temperature range of 600–900°C.

The formation of the L1₀(FePt) phase in the film with the copper sublayer is not established radiographically. However, the increase of the coercive force after annealing within the temperature range of 800–900°C can give evidence of the ordering process passing.

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REFERENCES

1. M. H. Kryder, *Proc of Symp. 'Computerworld's Storage Networking World Conference' (April 3–6, 2006)* (San Diego, CA, USA: Manchester Grand Hyatt: 2006), p. 350.
2. M. Albrecht and C. Brombacher, *phys. status solidi (a)*, **210**, Iss. 7: 1272 (2013).
3. O. P. Pavlova, T. I. Verbitska, I. A. Vladymyrskyi, S. I. Sidorenko, G. L. Katona, D. L. Beke, G. Beddies, M. Albrecht, and I. M. Makogon, *Appl. Surf. Sci.*, **266**: 100 (2013).
4. L. Liu, W. Sheng, J. Bai, J. Cao, Yu. Lou, Y. Wang, F. Wei, and J. Lu, *Appl. Surf. Sci.*, **258**: 8124 (2012).
5. K. Utsumiya, T. Seki, and K. Takanashi, *J. Appl. Phys.*, **110**: 103911 (2011).
6. <http://www.xakep.ru/magazine/xs/062/008/1.asp>.
7. K. Şendur and W. Challener, *Appl. Phys. Lett.*, **94**: 032503 (2009).
8. <http://www.idema.org/wp-content/downloads/1857.pdf>.
9. Y. S. Yu, Hai-Bo Li, W. L. Li et al., *J. Magn. Magn. Mater.*, **320**: L125 (2008).
10. Ch. Feng, Q. Zhan, B. Li et al., *Appl. Phys. Lett.*, **93**: 152513 (2008).
11. B. Wang, K. Barmak, and T. J. Klemmer, *IEEE Trans. Magn.*, **46**, No. 6: 1773 (2010).
12. W. Y. Zhang, H. Shima, F. Takano, H. Akinaga, X. Z. Yu, T. Hara, W. Z. Zhang, K. Kimoto, Y. Matsui, and S. Nimori, *J. Appl. Phys.*, **106**: 033907 (2009).
13. Iu. M. Makogon, E. P. Pavlova, S. I. Sidorenko, T. I. Verbitska, I. A. Vladymyrskyi, and R. A. Shkarban, *Metallofiz. Noveishie Tekhnol.*, **35**, No. 4: 553 (2013) (in Russian).
14. K. Barmak, J. Kim, L. H. Lewis et al., *J. Appl. Phys.*, **98**: 033904 (2005).
15. A. C. Sun, F. T. Yuan, and Jen-Hwa Hsu, *J. of Physics: Conference Series*, **200**: 1020099 (2010).
16. Yu. M. Makogon, O. P. Pavlova, S. I. Sidorenko, T. I. Verbitska, I. A. Vladymyrskyi, and O. V. Figurna, *Metallofiz. Noveishie Tekhnol.*, **35**, No. 10: 1425 (2013) (in Ukrainian).
17. T. Maeda, T. Kai, A. Kikitsu, T. Nagase, and J.-I. Akiyama, *Appl. Phys. Lett.*, **80**, No. 12: 2147 (2009).
18. J.-I. Ikemoto and Sh. Nakagawa, *J. Appl. Phys.*, **103**: 07B512 (2008).