Microstructure and Magnetic Properties of Iron Nitride Films Prepared by a Two-facing Targets Type DC Sputtering Apparatus

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Iron-nitride films were prepared by reactive sputtering with a two-facing targets type dc sputtering apparatus, and their magnetic properties as well as their film texture and crystal structure were investigated with reference to preparation conditions including sputtering conditions. The microstructure of films was observed by transmission electron microscopy supplemented with electron diffraction analysis. The magnetization curves were measured by a vibrating sample magnetometer (VSM), and $M_s$ and $H_c$ were evaluated in two directions parallel and perpendicular to the film plane. The magnetic domain structures were also studied by Lorentz electron microscopy. Various types of Fe-N compounds, $\varepsilon$-Fe$_2$N, $\gamma'$-Fe$_4$N and $\alpha$-Fe, were formed by controlling the sputtering conditions, and the phase stability of each phase was confirmed: A single phase $\gamma'$-Fe$_4$N film was prepared in two different ways; a direct deposition onto heated substrates and a successive heating of deposited films. Structural changes with annealing temperature are discussed with reference to the changes in magnetic properties. The change in magnetic coercivity with annealing temperature was found to be related to several parameters such as internal stress in film, grain size and precipitation process.

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I. Introduction

Recently, the authors(1)(2) have prepared iron films by a two-facing targets type dc sputtering apparatus, and demonstrated that the apparatus was suitable for the preparation of magnetic films because of a comparable high deposition rate of about 1 nm/s (600 Å/min). In the previous study the operation characteristic of the used apparatus suggested that it could be also used for various gas compositions. Therefore, in the present study, the apparatus was modified so as to prepare iron compound films by adding a reactive gas to a sputtering argon gas.

The preparation of iron nitride films has been carried out with various types of deposits including sputtering and evaporation. Kim and Takahashi(3) first prepared iron nitride films by evaporating iron in a nitrogen atmosphere, and reported that the iron atom has the largest magnetic moment in Fe$_{16}$N$_2$, which is not obtained by chemical reaction. Lo et al.(4)-(7) synthesized iron nitride films by rf sputtering using reactive sputtering gases composed of a mixture of argon and nitrogen containing a small amount of hydrogen, and investigated their crystal structure by X-ray diffractometry, microstructure by scanning electron microscopy and magnetic property by Mössbauer spectroscopy. They also demonstrated that discrete iron nitride films of $\gamma'$ and $\varepsilon$ phases, which are difficult to prepare by conventional chemical methods, were synthesized by controlling reactive sputtering conditions, and that the deposited film had a columnar structure with the columns perpendicular to the film plane. Also they(8) studied the chemical stability of the iron nitride films at elevated temperatures by means of Mössbauer spectroscopy in order to discuss the magnetic ordering in each phase of iron nitride films. Kano et al.(9) also prepared iron nitride films by reactive rf sputtering in a gas atmosphere containing nitrogen gas, and showed that the observed large saturation magnetization was attributed

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to Fe$_{16}$N$_2$ compound. Recently, Terada et al.\(^{(10)}\) have deposited iron nitride films using an ion beam deposition method, and showed that a film composed of multiphases such as $\alpha$-Fe, $\gamma'$-Fe$_4$N and unknown phases had a large magnetization $4\pi M_s$ of $2.40 \times 10^{-6}$ Wb/m$^2$ ($M_s = 1910$ G). They suggested that the iron nitride films with a large saturation magnetization and a low coercive force can be used as a magnetic layer of thin film type of magnetic head.

On the other hand, the $\gamma'$-Fe$_4$N compound film is expected as one of new recording media because of large saturation magnetization with a reasonable coercive force. From this point of view, Tazaki et al.\(^{(11)}\)(\(^{(12)}\)) reported that fine Fe$_4$N powder with a large magnetization and a coercive force (640 Oe) was prepared by nitrogenizing acicular metal iron powder. Umeda et al.\(^{(13)}\) prepared iron-nitride films by arc-discharge-type iron plating, and showed from the Mössbauer spectroscopy that the ferromagnetic films were highly corrosion-resistant.

Therefore, it is said that the iron nitride films are promising recording materials. However, in the above studies structural changes were studied by only X-ray diffractionmetry, and no microscopic study by electron microscopy was made with reference to the relation between the structural change and the film texture. In practice, film texture including grain size and the phase stability are very important to get appropriate magnetic properties such as the saturation magnetization $M_s$ and the coercive force $H_c$.

This study was carried out to clarify the magnetic properties of iron nitride films deposited by reactive dc sputtering. We attempted to synthesize various discrete iron nitrides and/or multi-phases by means of reactive sputtering with a two-facing targets type dc sputtering apparatus, and investigated the changes in film texture, constituent phase and magnetic properties with annealing temperature in vacuum. The relation between the microstructure and the magnetic property will be discussed on the basis of experimental results obtained.

### II. Experimental Procedure

Films were prepared in a two-facing targets type dc sputtering system, where a magnetic field of about 0.08 Wb/m$^2$ was applied perpendicularly to the target planes in order to avoid bombardment of ions on deposited films (Fig. 1). Two iron disks (purity; 99.99%), 100 mm in diameter and 5 mm thick, were used as target materials and placed to each other at the distance of 100 mm, and several substrates were placed perpendicular to and for away from the target surfaces, as shown in Fig. 1.

After the system was evacuated down to $2 \times 10^{-4}$ Pa, the sputtering gases of argon and nitrogen (both purity; 99.9999%) were introduced by handling needle valves for gas supply as well as a valve for high vacuum. In the present study, the partial nitrogen-gas pressure ($P_{N_2}$) was adjusted chiefly under a constant total pressure ($P_T$) of 0.6 Pa. Variable experimental parameters were: partial nitrogen-gas pressure of 0.02 to 0.4 Pa (discharge voltage of 710 to 780 V), discharge current up to 330 mA, sputtering time of $2 \times 10^2$ to $2 \times 10^3$ s and substrate temperature ($T_s$) of 293 to

![Fig. 1 Schematic diagram of a two-facing targets type dc sputtering apparatus.](image-url)
The used substrates were slide glass and 7059 glass for the measurements of magnetic property and crystal structure, and rock salt single crystal for the observation of film texture by means of electron diffraction and electron microscopy. The deposition rate was 15 to 25 nm/min (0.25–0.42 nm/s). The thickness of deposited films was measured using a Talystep profilometer. The sample films with various thicknesses were prepared under the same deposition conditions only by elongating the deposition time with manipulating a movable shutter. The crystal structure of as-deposited films was identified by X-ray diffractionmetry with CuKα and electron diffraction. The magnetization curves of films were measured using a vibrating sample magnetometer (VSM), and from the measured curves, the saturation magnetization Ms and the coercive force Hc were evaluated. The magnetic domain structures were also observed by means of Lorentz electron microscopy.

Furthermore, in order to study the phase stability of films, as-deposited films were annealed in a vacuum of 1.5 × 10⁻⁴ Pa at temperatures from 373 to 823 K (for 3.6 ks, and after cooled to room temperature their film texture was examined by electron microscopy. Some of them were in situ observed by an electron microscope equipped with a heating stage.

### III. Experimental Results and Discussion

#### 1. Crystal structure of produced Fe–N films

In the preliminary study the effect of the total sputtering gas pressure on the formation nature of iron nitrides was examined briefly. As a result, Fig. 2 shows typical relationships between deposition products and deposition conditions at various substrate temperatures. Here the crystal structure of each deposited film was identified at room temperature by X-ray diffraction. Therefore, each film examined was subjected to a successive cooling from a substrate temperature to room temperature.

![Fig. 2 Relation between the constituent phase of as-deposited films and the deposition conditions including sputtering conditions](image)

| Substrate Temperature (K) | P_N/P_T=0.15 | P_T=0.6 Pa |
|---------------------------|--------------|-----------|
| ε                         | e + γ'       | γ'        | 573 K |
| e + γ'                    | γ'           |          | 473 K |
| e                         | e + γ'       | R.T.      |

| Substrate Temperature (K) | P_N/P_T=0.15 | P_T=0.4 Pa |
|---------------------------|--------------|-----------|
| ε + γ'                    | γ'           | 573 K     |
| e + γ'                    | γ'           | 473 K     |
| e                         | e + γ'       | R.T.      |

| Substrate Temperature (K) | Tₛ=473 K | P_T=0.6 Pa |
|---------------------------|---------|-----------|
| γ + α + (e)               | α + γ'  | 0.15      |

One can see from Fig. 2 that various types of Fe–N compounds are produced by controlling the sputtering conditions and substrate temperatures under various gas pressures. A comparison between Fig. 2(a) and (b) suggests that the total gas pressure and the substrate temperature have opposite effects on the formation of single ε-phase nitrides: When the ratio P_Total/P_T is constant, an increment of the total gas pressure results in an increase in number of nitrogen atoms in the reactive atmosphere and, thus, the single ε-phase nitrides containing much nitrogen in films are formed in a wide discharge current range with increasing the total gas pressure, while the formation range.

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The deposition rate of this apparatus depends on the applied magnetic field, and, thus, may be improved by increasing the applied field.
for the ε-phase nitrides becomes narrow with increasing substrate temperature. The increase in substrate temperature may cause the decomposition of nitrides or the extraction of nitrogen atoms from nitride films. The nature of the formation of each nitride indicates that the nitrogen content in deposited films decreases with increasing discharge current. This can be explained by considering that, when the discharge current increases, sputtered iron atoms increase in number, while the number of iron atoms which can react on nitrogen atoms is restricted by a given partial nitrogen-gas pressure. It was also found that there was no evidence of the formation of the Fe₆₆N₂ phase for all films examined.

Figure 3 reproduces a series of electron micrographs showing a structural change during heating for a thin film which was deposited on a rock salt substrate at room temperature.

Fig. 3 A series of electron micrographs showing a structural change during heating in a thin film deposited onto a rock salt substrate at room temperature under $P_{N_2}=0.073$ Pa (0.55 mTorr, $P_{N_2}/P_1=0.11$) with a discharge current of 220 mA.
under a partial nitrogen-gas pressure of 0.073 Pa (0.55 mTorr, $P_{N2}/P_t=0.11$) with a discharge current of 220 mA. The as-deposited film was identified to be $\varepsilon$-Fe$_2$$_3$N, and the diffraction pattern inserted in the figure appears broad, indicating that the as-deposited film is composed of many small grains and strained heavily probably by a large number of crystal defects. At 623 K, $\gamma'$-Fe$_4$N begins to precipitate along the crystal boundaries of $\varepsilon$-Fe$_2$$_3$N grains, and then $\gamma'$-Fe$_4$N grains grow with increasing annealing temperature until the dissolution of the $\varepsilon$-Fe$_2$$_3$N phase is completed. Similar in situ observations were carried out up to elevated temperatures for deposited films which were prepared under different partial nitrogen-gas pressures, and the results are summarized in Fig. 4. A similar characteristic of phase stability is observed for all the films examined: All deposited films have an $\varepsilon$-Fe$_2$$_3$N structure in a low temperature range, the $\varepsilon$-phase is stable below a critical temperature, and above this temperature the $\gamma'$-phase precipitates from the $\varepsilon$-phase. The single-phase film of $\gamma'$-Fe$_4$N is obtained when annealed between 723 and 823 K; the $\gamma'$-phase is stable up to 823 K. Above 823 K, the $\gamma'$-phase dissolves so fast that we cannot observe how the $\alpha$-Fe is formed from the $\gamma'$-phase. The films, when annealed above 823 K for 72 ks, show a mixture of $\alpha$-Fe and (fcc) phases. This (fcc) compound, which has not been identified yet, precipitates along the grain boundaries of the $\gamma'$-phase with a disk shape of 20 nm diameter.$^{(14)}$ The lattice parameter of this compound was 0.84 nm. $^{(15)}$ have already reported that a similar Fe-N compound with fcc phase was observed when an iron containing nitrogen was heated at an elevated temperature more than 600 K. Further a detailed study of the precipitation process from $\varepsilon$-phase to the $\gamma'$-phase is in progress and the results will be reported elsewhere.$^{(14)}$

A comparison between Fig. 2 and Fig. 4 shows that the stability of each nitride phase is essentially similar irrespectively of heating conditions imposed on the films and, thus, suggests that various kinds of iron nitride phases including discrete phases are formed by controlling the sputtering conditions and by the annealing conditions: for example, using the present two-facing targets type dc sputtering apparatus a single-phase $\gamma'$-Fe$_4$N is prepared in two different ways: one is by a direct sputtering onto heated substrates under selected sputtering conditions and other is by an annealing of as-deposited films at a suitable annealing temperature.

As was expected, an X-ray analysis supplemented by electron microscopy showed that the texture of as-deposited films was very different from that of annealed films, although each experimental result was not given here: The as-deposited film has a partially preferred [001] orientation with a small grain size, while the annealed one, composed of recrystallized grains, has no preferred orientation.

2. Magnetic properties of Fe-N films

Figure 5 shows typical magnetization curves of as-deposited films which have a single-phase
(or a nearly single-phase) of $\varepsilon$, $\gamma'$ and $\alpha$, respectively: the symbols $\parallel$ and $\perp$ mean the directions of applied magnetic field parallel and perpendicular to the film surface, respectively. The saturation magnetization $4\pi M_s$ of each film is comparably large, and especially $\alpha$-Fe ($2.32 \times 10^{-6}$ Wb/m$^2$) has a larger magnetization than pure iron ($2.187 \times 10^{-6}$ Wb/m$^2$). This suggests that the iron nitride film is suitable for a magnetic layer type of magnetic head for high-density recording, although the stability for pyrophoric circumstance of the film must be established.

Figure 6 shows an influence of annealing temperature on the coercive force of Fe-N films. The films were prepared at normal temperature under deposition conditions noted in each diagram, annealed at each elevated temperature for 3.6 ks, and subsequently cooled to room temperature for the measurement of coercive force. The corresponding crystal structure is shown at the upper part of each figure. For a film deposited on 7059 glass (Fig. 6(b)), the coercive force first decreases and subsequently increases with increasing annealing temperature, and a minimum of coercive force is observed around 623 K. Similar changing modes were observed for the other films shown in Fig. 6(a) and (c), although the absolute value of coercive force was different from each other. This experimental finding, however, is not explained in terms of only constituent phases of the film, because each film has different constituent phases. Accordingly, in order to explain the above changing mode of coercive force, alternative factors must be taken into consideration.

Figure 7 shows changes in the half-value width of 002 reflection peak for the $\gamma'$ phase in the films (a) and (b) and of the 110 reflection peak for the $\alpha$-phase in the film (c), respectively, as a function of annealing temperature. The films used here are the same as those used for the measurement of coercive force in Fig. 6. The half-value width decreases with annealing temperature and finally reaches almost the

![Figure 5](image-url)  
Fig. 5 Magnetization (H–M) curves of iron nitride films. Each phase of film was confirmed by X-ray analysis, in which the main peak was used for identification of crystal phase. Film thickness: 1.15 μm. (a) $\varepsilon$–Fe$_{2.3}$N, (b), $\gamma'$–Fe$_4$N, (c), $\alpha$–Fe.

![Figure 6](image-url)  
Fig. 6 An influence of annealing temperature on the coercive force of Fe-N films. Deposition conditions for each film are noted in each diagram.
same value for each film after annealing at a higher temperature. As is well known, the half-value width reflects both a grain size distribution and a strain state of material. For the present experiments where the films examined were prepared at the same time under the same deposition conditions, a film texture as well as grain size is considered to be almost the same for both film (a) and film (b). Therefore, the half-value width reflects mainly a difference in stress state of film.

Recently, we have investigated the internal stresses in such deposited films by X-ray diffraction techniques and demonstrated that the deposited films were in a state of mechanical stress and such a stress including a film stress due to the difference in thermal expansion was easily released by annealing. Thus, the above changes in half-value width indicate that the internal stress is reduced more easily in the film (b) than in the film (a). This is consistent with Kersten's proposal that a heavily strained material (or film) is easier to recrystallize in order to release its internal stress. Therefore we can say that the internal stress in a little larger in the film (b) than in the film (a). Kersten has proposed that the larger the internal stress becomes the higher the coercive force is. According to his concept, the present result is explicable on the basis of the fluctuations on internal stress in film. Furthermore, a comparison between Fig. 6(a) and (c) may suggest that the effect of internal stress on the magnetic properties varies with constituent phases which have different crystal structures and crystal symmetry, although the influence of film texture is not neglected. The observed changing mode of coercive force with increasing annealing temperature will be discussed in the next section III.4.

3. Magnetic domain structure of Fe–N films

The above-mentioned changes in the magnetic properties are considered to be related to the changes in magnetic domain structure. Therefore, in situ observations of the magnetic domain structure were made by means of Lorentz electron microscopy. Figure 8 shows a series of Lorentz electron micrographs showing a change in magnetic domain structure with annealing temperature. A film examined was deposited on a rock salt crystal at room temperature under the following conditions: applied voltage of 750 V, discharge current of 220 mA and $P_{N_2}$ of 0.063 Pa. After being peeled off out of the substrate by dipping in water, the film was installed in an electron microscope equipped with a heating stage, and successive observations were made. Another film which was prepared onto the 7059 glass under the same deposition conditions with a different thickness was examined by VSM, and the change in coercive force corresponding to each annealing temperature is also shown schematically in a diagram inserted in Fig. 8. Figure 8(a), taken in an as-deposited state, shows a typical example of magnetic do-
Fig. 8  A series of Lorentz electron micrographs showing a change in magnetic domain structure with annealing temperature. Each measuring temperature where each electron micrograph was taken is indicated by the arrows in the inserted diagram in the right bottom.
main structure: divergent (dark) and convergent (white) images of main magnetic walls indicated by the arrows, and a set of curved fine linear images intersecting the magnetic walls (the so-called ripple structure). Several inclined banded images having a low contrast are due to steps on the rock salt substrate. It has been reported that the ripple line appears to be perpendicular to the mean magnetization direction and, thus, the ripple structure reflects a magnetization state in film\(^{(10)}\). The geometrical feature of the observed ripple structure indicates that there is a small fluctuation of magnetization direction within the film plane. This is consistent with the experimental finding that the as-deposited films have a partially preferred orientation, as described in the section III.1. When the annealing temperature was raised, the main domain walls showed no remarkable change in image contrast as well as in their position, while the ripple lines changed their image contrast together with the appearance of branches off the main ripple lines, and finally diminishes their initial line shape (see Fig. 8(c) to (e)). The change in the magnetic domain structure may be related with the following structural change of film: the recrystallization process takes place at 473 K and recrystallized grains grow large with increasing annealing temperature, followed by a diminution of preferred orientation in film texture. As a result, the mean magnetization direction in film has some dispersion around the initial directions. Above 673 K the crystal grains continue growing and, furthermore, the precipitation of \(\gamma'\)-Fe\(_{2}\)N from \(\varepsilon\)-Fe\(_{2}\)\(_{3}\)N proceeds with increasing annealing temperature, and finally the initial domain structure disappears completely around the Curie temperature of \(\gamma'\)-phase (Fig. 8(g)) (around 760 K\(^{(12)}\)).

4. Origins of coercive force change

Here, we discuss the coercive force change with annealing temperature. Generally, the coercive force of ferro-magnetic materials is discussed on the basis of internal stress\(^{(16)}\), crystal structure, grain size\(^{(19)}\), and local magnetic anisotropy including shape anisotropy\(^{(20)}\). When the Fe-N films were annealed up to 623 K, several changes were observed together with a decrease of coercive force: the constituent phases of film did not change, but the corresponding magnetic domain structure changed accompanying changes in the ripple structure. The half-value width also decreased with the annealing temperature. According to Kersten's concept\(^{(16)}\), the relation between the coercive force \(H_c\) and the internal film stress \(\sigma\) is given approximately by \(H_c = A\sigma\), where \(A\) is a constant. Since the internal stresses introduced on deposition process are released by annealing, a relaxation of internal film stress may result in a decrease in coercive force. On the contrary, when the annealing temperature was above 623 K, the internal stress was released almost completely and, thus, it would not give any further influence on the coercive force. Main changes occurring in the films are: the grain growth accompanied with internal stress relaxation due to recrystallization, and the precipitation of \(\gamma'\)-Fe\(_{2}\)N from \(\varepsilon\)-Fe\(_{2}\)\(_{3}\)N accompanied with the precipitation of \(\alpha\)-Fe. It is well known that the coercive force increases with increasing grain size\(^{(19)}\). Furthermore, an existence of a different phase along the grain boundaries of the matrix phase will give rise to an increase in anisotropy energy of magnetization\(^{(20)}\) and, thus, the precipitates can play a role of obstruction to the motion of magnetic domain walls (the so-called pinning effect). These also result in an increase in coercive force. Therefore, the combination of the above processes may influence a changing mode of coercive force, and as a result the coercive force increases with increasing annealing temperature. Further detailed studies must be made to understand the origin of the coercive force change.

IV. Conclusions

Iron-nitride films were prepared by using nitrogen gas as a part of sputtering gas in a two-facing targets type dc sputtering system, and their magnetic properties as well as their crystal structures were investigated with reference to the preparation conditions including heat-treatments. The main results are
summarized as follows:

1) The film structure was very sensitive to deposition conditions such as the substrate temperature, partial nitrogen-gas pressure and discharge current. This indicates that we can obtain iron nitride films which have an intended crystal structure: The single-phase $\gamma'$-Fe$_4$N film was synthesized by a direct sputtering method and also by a successive heat-treatment after deposition. The iron nitride film will be used as a magnetic layer of thin film type of magnetic head for high density recording and as a core material of inductor used at high frequency.

2) The changes in film texture with annealing temperature are as follows: As-deposited films are composed of small crystallites accompanied with internal stresses, and they have a partially preferred orientation in film texture. Recrystallization process takes place at 473 K without any remarkable change in grain size, and the internal stresses appear to be released completely above 623 K. Above 523 K, $\gamma'$-Fe$_4$N phase precipitates from the $\varepsilon$-Fe$_{2.3}$N phase and the $\gamma'$-Fe$_4$N phase is stable up to 823 K. Consequently, the preferred orientation in film texture disappeared after recrystallization and precipitation processes.

3) Corresponding changes in magnetic domain structure are as follows: Most of as-deposited films have the mean magnetization direction in the film plane. At the beginning of annealing treatment, a main magnetic domain structure does not change up to 623 K, but a slight change in magnetization orientation was observed as a change in ripple structure. After recrystallization, large changes were recognized in magnetic domain structure, indicating that the magnetization direction was dispersed within the film.

4) Corresponding changes in magnetic coercive force are as follows: With increasing annealing temperature, the coercive force of films decreases and has a minimum value at 623 K (the first stage), and then increases again (the second stage). The first stage is attributed mainly to the diminution of internal stresses induced in film, although an influence due to a change in film texture is not negligible, and the second stage is due to the crystal grain growth, accompanied by the precipitation processes of $\gamma'$-Fe$_4$N from $\varepsilon$-Fe$_{2.3}$N and of $\alpha$-Fe from $\gamma'$-Fe$_4$N.

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REFERENCES

(1) S. K. Gong and O. Nittono: J. Japan Inst. Metals, 50 (1986), 935.
(2) T. Kaneko, S. K. Gong and O. Nittono: Trans. JIM, 27 (1986), 241.
(3) T. K. Kim and M. Takahashi: Appl. Phys. Lett., 20 (1972), 492.
(4) Cary Lo, S. V. Krishnaswamy, K. R. P. M. Rao, R. Messier and L. N. Mulay: Mat. Res. Bull., 15 (1980), 1267.
(5) C. Lo, S. V. Krishnaswamy, L. N. Mulay and D. A. Diffenbach: J. Appl. Phys., 53 (1982), 2745.
(6) C. Lo and L. N. Mulay: Mat. Res. Bull., 19 (1984), 261.
(7) C. Lo, S. V. Krishnaswamy, R. Messier, K. R. P. M. Rao and L. N. Mulay: J. Vac. Sci. Technol., 18 (1981), 313.
(8) C. Lo and L. N. Mulay: Mat. Res. Bull., 18 (1983), 1425.
(9) A. Kano, N. S. Kazama, N. Fujimori and T. Takahashi: J. Appl. Phys., 53 (1982), 8332.
(10) N. Terada, Y. Hoshi, M. Naoe and S. Yamanaka: IEEE Trans. Magn., MAG-20 (1984), 1451.
(11) A. Tasaki, K. Tagawa, E. Kita, S. Harada and T. Kusunose: IEEE Trans. Magn. MAG17 (1981), 3026.
(12) K. Tagawa, E. Kita and A. Tasaki: Jpn. J. Appl. Phys., 21 (1982), 1596.
(13) K. Umeda, Y. Kawashimo, M. Nakasone, S. Harada and A. Tasaki: Jpn. J. Appl. Phys., 23 (1984), 1576.
(14) S. K. Gong, Y. Nakamura and O. Nittono: Collected Abstracts of the 1986 Autumn Meeting of Japan Inst. Metals, p. 223.
(15) Жигалов В С, Вершина Л И and Фролов Г И: Fiz. Tverd. Tela., 26 (1984), 1887.
(16) M. Kersten: Phys. Z. 39 (1938), 860.
(17) O. Nittono, Y. Sadamoto and S. K. Gong: Jpn. J. Appl. Phys., 26 (1987) 157.
(18) M. W. Fuler and M. E. Hale: J. Appl. Phys., 31 (1960), 238.
(19) H. Hoffmann: IEEE Trans. Magn. MAG-9 (1973), 17.
(20) H. Fujimori, H. Morita, Y. Obi and S. Ohta: Amorphous Magnetism, eds. R. A. Levy and R. Hasegawa, Plenum Press, New York, (1977), p. 393.