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The influence of the textural development on magnetic properties in Co–Cr thin films

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Abstract

Two series $Co_{81}Cr_{19}$ films (5–980 nm) were chosen to investigate the influence of textural development on magnetic properties. It was demonstrated that both the texture and the magnetic properties strongly depend on the film thickness. In the thickness range from 12 to 980 nm, the coercivity $H_{c\perp}$ and the orientation ratio $OR_{m\perp}$ of the magnetization are mainly determined by the orientation of a hcp crystallites. In the case of Series B films having thickness range of 46–980 nm, there is a critical thickness, at which the $H_{c\perp}$, the orientation ratio $OR_{m\perp}$ and the orientation ratio OR_c of the crystallite reach their maximum values and the strain ϵ in film is about zero. As the thickness increases, the magnetostriction anisotropy constant gradually changes from the negative to the positive value. The changes of the calculated $H_{c\perp}$ with the thickness agree much better with experimental value if the magnetostatic interaction is considered in calculation. Without considering the effect of the magnetostatic interaction, the agreement with experiment is poor. This result proves that $H_{c\perp}$ is mainly controlled by the magnetocrystalline and shape anisotropy, and that the magnetostriction anisotropy exerts a rather small influence on $H_{c\perp}$. In the thickness range of 5–12 nm, the $Co_{81}Cr_{19}$ films exhibit an anomalous behaviour, and the observations can be made: (1) $H_{c\perp}$ is almost independent of the thickness; (2) the squareness ratio $R_{s\parallel}$ increases with increasing thickness, even though the orientation ratio OR_c of the crystallite increases with increasing thickness.

1. Introduction

Co–Cr sputtered films have been extensively investigated as perpendicular recording media where the microstructure can be developed with a columnar grains and a strong texture [1–8]. A vast amount of experimental data shows that the growth direction of the columns and the (0002) direction of a hcp crystallite are oriented along the direction normal to the film. Obviously, such microstructure favours the alignment of the magnetization in the direction normal to the film plane. As was demonstrated in our previous paper [9], for two series of $Co_{81}Cr_{19}$ films having the thickness range between 5 and 980 nm, the texture and the internal strains change with the film thickness [Tables 1 and 2]. For Series B films, there is a critical film thickness, below which the OR_c value increases with increasing thickness

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and above which the OR_c value decreases with increasing thickness. In the vicinity of 130 nm, the orientation ratio OR_c reaches its maximum value. The strain changes with the film thickness as well, as the film thickness increases from 46 to 980 nm, the film strain e along the film normal gradually changes from a tensile to compressive strain, and in the vicinity of 130 nm the film is stress-free. Therefore, for Series B films both the highest OR_c and the lowest e along the film normal appear at the same thickness. In this case, a high quality perpendicular recording media with strong texture and stress-free conditions can be prepared by optimizing the film thickness. In addition, as the film thickness increases from 5 to 980 nm, the aspect ratio (the ratio of the column length along the film normal to the columnar diameter) gradually increases from about 1 to 7. It means that as the film thickness increases, the shape of the grains gradually changes from the equiaxial shape to the typical columnar one. However, even for the thinnest film (5 nm), where the aspect ratio approaches 1, the (0002) texture is rather strong, 3.5 times stronger than that in a specimen without texture.

Based on the above results, the relationship between the texture and magnetic properties will be investigated in this paper and we will focus on the dependence of the magnetic properties on the texture and the thickness of $Co_{81}Cr_{19}$ films.

2. Experimental procedure

Two series of sputtered $Co_{81}Cr_{19}$ films were chosen to investigate the correlation between the texture evolution and the magnetic properties, which is affected by changes in the film thickness. Series A and B films were sputtered on the Ge/SiO_2/Si and Si substrate respectively. The Si substrate is a single crystal of Silicon and the substrate plane coincides with [100] planes. The thickness of the films ranges from 5 to 980 nm. To measure the thickness the SEM, STM (Scanning Tunnel Microscope) and/or AFM (Atomic Force Microscope) were used. The magnetic properties were measured by VSM and a Torque magnetometer. The maximum applied field for measuring the hysteresis loop is about 840 kA/m. The most relevant properties for Series A and B films are summarized in Tables 1 and 2 respectively, where the $R_{s\perp}$ and $R_{s\parallel}$ are the squareness ratio $M_r \perp / M_s$ and $M_r \parallel / M_s$, along perpendicular and in-plane direction respectively. $OR_{m\perp}$ is the so-called orientation ratio of the magnetization and is defined as the ratio of $R_{s\perp}/R_{s\parallel}$, reflecting the anisotropy of the remanence along the easy and hard direction. Obviously, to achieve a high quality perpendicular recording medium, it is desirable to have the $OR_{m\perp}$ value as high as possible and the $R_{s\parallel}$ value as low as possible.

In order to investigate the texture changes with the thickness, all the specimens were first analyzed using Rigaku diffractometer (Cu-Kα radiation) to determine the crystalline structure, then the texture of the specimens was measured using a Siemens D-500 texture goniometer (Mo-Kα radiation). The pole figures for hcp (0002) crystallographic planes were then plotted. In addition, the precise orientation distribution of (0002) pole density was measured. This pole density F(α) was registered as a function of angular deviation from the film normal. Data were col-

Table 1
Properties for Series A $Co_{81}Cr_{19}$ films

No.	Thickness nm	M_s kA/m	$H_{c\perp}$ kA/m	K_1 KJ/m^3	$R_{s\parallel}$	$OR_{m\perp}$	OR_c	$\Delta\alpha_{50}$
A_1	5	360	4.0	122	0.08	12.5	3.5	4.78°
A_2	12	460	4.3	116	0.83	0.02	9.6	3.9°
A_3	25	500	15.4	127	0.33	0.18	22	3.3°
A_4	50	485	36.8	112	0.15	0.73	32	2.8°
A_5	100	510	47.2	134	0.09	1.34	40	2.9°
A_6	200	500	69.9	131	0.06	2.84	44	3.0°

Table 2
Properties for Series B $Co_{81}Cr_{19}$

No.	Thickness nm	M_s kA/m	$H_{c\perp}$ kA/m	K_1 KJ/m^3	$R_{s\parallel}$	$OR_{m\perp}$	OR_c	$\Delta\alpha_{50}$
B_1	46	450	79	94	0.11	2.4	31	3.0°
B_2	110	470	96	115	0.03	7.5	41	3.0°
B_3	297	445	65	99	0.03	5.1	36	3.5°
B_4	611	435	46	84	0.03	3.2	27	4.0°
B_5	982	430	40	86	0.04	2.7	19	4.5°

lected in a 12° angular interval as a function of the specimen tilt angle α. The ratio $OR_c(\alpha,\beta)$ of the pole density $F(\alpha,\beta)$ of the film to that of a random specimen was used to evaluate the orientation ratio using the following expression:

$$OR_c(\alpha,\beta) = \frac{\left(\frac{1}{\alpha_2-\alpha_1}\right)\int_{\alpha_1}^{\alpha_2}\int_0^{2\pi} F(\alpha,\beta)\,d\alpha\,d\beta}{\frac{2}{\pi}\int_0^{\frac{\pi}{2}}\int_0^{2\pi} F(\alpha,\beta)\,d\alpha\,d\beta} \tag{1}$$

where β is the circumferential angle of pole figure. In our case, the value of $F(\alpha,\beta)$ is assumed to be equal to $F(\alpha)$, since the texture in the α-Co phase exhibits high degree of rotational symmetry about the film normal [9]. The difference $(\alpha_2-\alpha_1)$ is equal to 0.1°. For the sake of convenience, in this report the OR_c value, which represents the maximum $OR_c(\alpha,\beta)$ for (0002) planes, will be called the orientation ratio of the crystallite. The $\Delta\alpha_{50}$ FWHM (full width-half maximum), which is derived from the precise orientation distribution curve for (0002) planes, will be used to characterize the dispersion of (0002) planes.

Based on the model discussed earlier [11], where the magnetic material is considered as an assembly of the single domain fine particles, and we assume the Co–Cr film is composed of the single domain grains, which are completely isolated and separated from each other by paramagnetic Cr-rich boundaries, the coercivity $H_{c\perp}$ can be calculated using the following formulae:

$$H_{c\perp} = \frac{2T}{M}\sum_{L=0}^{L} T_L\delta_{CL} \tag{2}$$

$$\frac{2T}{M} = H_a + H_d + gH_\sigma$$
$$= \frac{2K_1}{M_s} + (N_0 - N_t)M_s + g\frac{3\sigma\lambda}{M_s}. \tag{3}$$

Coefficient δ_{CL} is mathematical constant representing the texture influence on the coercive force. This coefficient can be found out in references [6,11]. The term of $2T/M$ is, in fact, the effective anisotropy field. The terms H_a, H_d and H_σ in the formula (3) represent the magnetocrystalline, the shape and the stress anisotropy field. N_0 and N_t are the demagnetizing factor along the easy and the hard direction respectively. K is the magnetocrystalline anisotropy constant. In our case, K is replaced by the first order anisotropy constant K_1, because for most of Co–Cr films the second order anisotropy constant is much smaller than K_1. σ and λ_s are the stress and the saturation magnetostriction coefficient ($= -12\times10^{-6}$) in Co–Cr film respectively. g is a fitting coefficient.

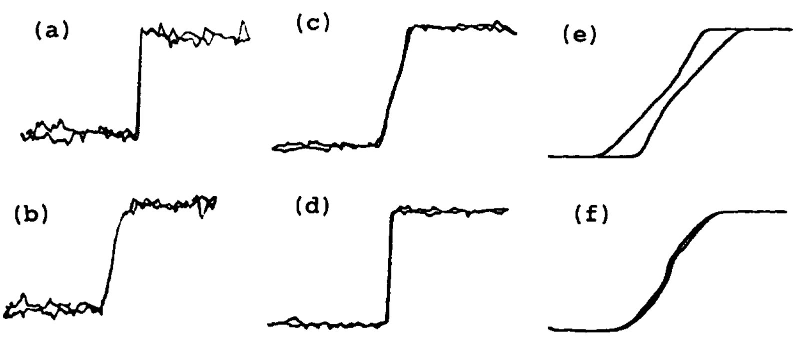

Fig. 1. Typical VSM hysteresis loops measured for different $Co_{81}Cr_{19}$ films. (a) 5 nm, along the perpendicular direction; (b) 5 nm, along the in-plane direction; (c) 12 nm, along the perpendicular direction; (d) 12 nm, along the in-plane direction; (e) 100 nm, along the perpendicular direction; (f) 100 nm, along the in-plane direction.

The related texture expansion coefficients T_L in formula (2) are calculated using the formula:

$$T_L = \frac{(2L+1)}{(2)} \int_{\alpha_1}^{\alpha_2} F(\alpha) P_L(\alpha)\, \mathrm{d}\alpha = \frac{(2L+1)}{2} \sum_{L=0}^{L} \sum_{\alpha_1}^{\alpha_2} F(\alpha) P_L(\alpha) \Delta\alpha. \quad (4)$$

$F(\alpha)$ is the measured orientation distribution curve of (0002) planes and $P_L(\alpha)$ is the Lth order Legendre polynomial. Considering the fact that this orientation distribution curve $F(\alpha)$ for tested $Co_{81}Cr_{19}$ films exhibits a symmetrical shape in the vicinity of the film normal, only the Legendre polynomials having even L values are involved during the calculation of $H_{c\perp}$ and $M_r \perp / M_s$. In our case, the maximum L value of the series expansion is taken as 20.

It is necessary to note that the influence of the film thickness on $H_{c\perp}$ is implicitly included in the calculation of the demagnetizing field and the coefficients T_L in the formulae (2), (3) and (4), because the measured orientation distribution $F(\alpha)$ of (0002) planes and the demagnetizing factor depend on the film thickness.

3. Results and discussion

3.1. Magnetic property in films having different thicknesses

The typical measured VSM hysteresis loops for Series A films having thicknesses of 5, 12 and 100 nm are shown in Fig. 1. These hysteresis loops can be classified into three types: (a) for the thinnest film A_1 (5 nm), the hysteresis loop measured in direction perpendicular to the film has rectangular shape, but the hysteresis loop along in-plane direction exhibits very low $R_{s\parallel}$ value of 0.08. It indicates that the magnetization is perfectly oriented along the film normal; (b) For film A_2 (12 nm), the hysteresis loop along in-plane direction has rectangular shape, but the hysteresis loop along perpendicular direction has lower $R_{s\parallel}$ value of 0.02. It suggests that the magnetization is aligned in-plane direction; (c) For film A_5 having thickness of 100 nm, the hysteresis loop displays the behaviour characteristic for a perpendicular recording media, except that there exists a 'shoulder' on the knee part of hysteresis loop. Experimental data show that for all other films having thickness higher than 12 nm, the shape of the hysteresis loop is similar to that for film having thickness of 100 nm, but the 'shoulder' move to a higher field as the thickness increases, indicating that the magnetization reversal is partly dictated by the domain-wall motion [12]. In addition, for film A_3 (25 nm) a 'jump' on the in-plane hysteresis loop was observed in the vicinity of zero applied field. Such a jump in the magnetization of the in-plane hysteresis loop in the vicinity of zero field was also observed earlier [12,13].

Based on described changes in the characteristics of the hysteresis loops with thickness, the film thickness of 12 nm should be considered as being a transition point. In the vicinity of 12 nm, the $Co_{81}Cr_{19}$ film exhibits the behaviour observed for magnetization oriented along longitudinal direction. For other films, whose thickness deviate from this critical value, no matter how thin or thick the film is, the magnetization is always aligned along the film normal with a varying degrees of alignment.

In principle, for a high quality perpendicular recording medium, the higher the effective anisotropy in perpendicular direction, the larger is the coercivity $H_{c\perp}$ and the higher the orientation ratio $OR_{m\perp}$ will be. In the same film, the smaller the squareness ratio $R_{s\parallel}$ and the coercivity $H_{c\parallel}$ along in-plane direction should be. For both Series A and B films having the thickness higher than 12 nm, as shown in Tables 1 and 2, as the film thickness increases, a pattern of magnetic properties changes with thickness exhibits the characteristics, which are typical for perpendicular recording media. For example, for Series A films, as the thickness increases from 12 to 200 nm, the coercivity $H_{c\perp}$ increases from 4.3 to 69.9 kA/m and the orientation ratio $OR_{m\perp}$ increases from 0.024 to 2.8. For the same specimens the squareness ratio $R_{s\parallel}$ along the in-plane direction decreases from 0.83 to 0.06.

It was demonstrated earlier [9] that in the same thickness range, for Series A films the orientation ratio OR_c of the crystallite increases with increasing thickness. Based on these experimental data, it can be argued that for $Co_{81}Cr_{19}$ films having thickness higher than 12 nm, the magnetic properties are mainly determined by the magnetocrystalline anisotropy, because both $OR_{c\perp}$ and $OR_{m\perp}$ increase with increasing OR_c value. However, if the film thickness is lower than 12 nm, the dependence of magnetic properties on the thickness exhibits anomalous characteristics. It is evident from Table 1 that as the thickness increases from 5 to 12 nm, the orientation ratio OR_c increases from 3.5 to 9.6, but the coercivity $H_{c\perp}$ is almost independent of the thickness and the orientation ratio $OR_{m\perp}$ decreases from 12.5 to 0.02. The squareness ratio $R_{s\parallel}$ along the in-plane direction, at the same time, increases from 0.08 to 0.83.

In addition, the normalized hysteresis loss as a function of the applied field measured for different film thicknesses for Series A films can be also used to prove the anomalous magnetic properties of films. The normalized hysteresis loss as a function of the applied field for Series A films was plotted in Fig. 2. In this figure we observe that for 5 nm film, the normalized hysteresis loss decreases monotonically with increasing of the applied field. Obviously, this pattern is quite different from those for other films, whose thicknesses are higher than 12 nm. In addition, for Series A films having thickness higher than 12 nm, the curve representing the normalized hysteresis loss $\Delta W_{h\perp}$ vs. applied field have a single peak only, and the peak position always deviates from the zero field. As the thickness increases, the peak position gradually moves to a higher applied field and the magnitude of these peaks decreases monotonically.

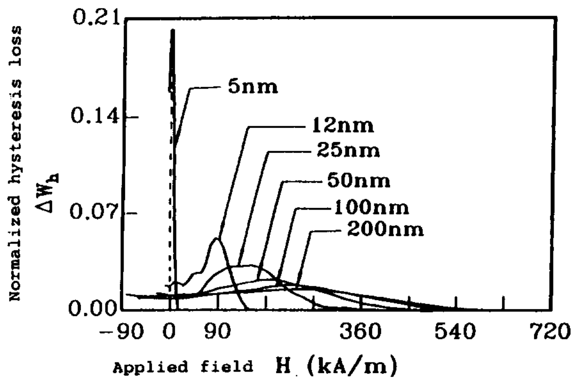

Fig. 2. The normalized hysteresis loss as a function of the applied field with the thickness as a parameter for Series A films.

Summing up, the anomalous relationship between the magnetic properties and the texture for $Co_{81}Cr_{19}$ films, for thicknesses less than 12 nm, can be expressed by the following observations:

(a) $H_{c\perp}$ is almost independent of the thickness.

(b) The orientation ratio $OR_{m\perp}$ decreases with increasing the thickness.

(c) The squareness ratio $R_{s\parallel}$ increases with increasing thickness.

(d) The peak of the hysteresis loss vs. applied field appears at magnetic field close to zero, even though the orientation ratio OR_c of the crystallite still increases with increasing thickness. This means that if the thickness increases from 5 to 12 nm, the structure of the $Co_{81}Cr_{19}$ film changes from a typical perpendicular anisotropy media to a longitudinal one upon condition that the texture of (0002) planes does increase in strength.

On the other hand, for Series B films, as is shown in Table 2, as the film thickness increases, the orientation ratio $OR_{m\perp}$ and also the coercivity $H_{c\perp}$ first increase with increasing the thickness, and then decrease with increasing thickness. It can be concluded from these observations that there is an optimum thickness, where the $OR_{m\perp}$ and $H_{c\perp}$ values reach the maximum values. For example, as the thickness increases from 46 to 110 nm, the $OR_{m\perp}$ value increases from 2.4 to 7.5 and $H_{c\perp}$ from 79 to 96 kA/m. On the contrary, if the thickness continues to increase from 110 to 980 nm, the $OR_{m\perp}$ and $H_{c\perp}$ values monotonically decrease from 7.5 to 2.7 and from 96 to 40 kA/m respectively. In addition, a similar pattern of the change in the dependence of OR_c on the thickness was also observed.

Combining the change in the pattern of the orientation ratio OR_c with those of the abovementioned magnetic properties, it can be concluded that for films thicker than 12 nm, the magnetocrystalline anisotropy exerts a dominant influence on the magnetic properties.

3.2. The relation between the measured magnetic properties and the texture

In order to illustrate the influence of the textural evolution on the magnetic properties of specimens having different thicknesses, the relation between the measured orientation ratio OR_c and the measured orientation ratio $OR_{m\perp}$ as a function of the thickness was plotted in Fig. 3. It can be clearly seen from this figure that for Series A films having thicknesses higher than 12 nm, as the thickness increases from 12 to about 200 nm, both $OR_{m\perp}$ and OR_c values increase with increasing thickness. On the other hand, for Series B films. Both $OR_{m\perp}$ and OR_c values first increase with increasing thickness, then their values decrease with increasing thickness if the film thickness exceeds the optimum thickness. In the vicinity of 110 nm, both OR_c and $OR_{m\perp}$ values reach their maximum values. This means that the changes of both OR_c and $OR_{m\perp}$ values with thickness follow a similar variation. Comparing Series A films with Series B films, Series B films have larger $OR_{m\perp}$ values than those of Series A films, indicating that the substrate exerts significant influence on the $OR_{m\perp}$ value.

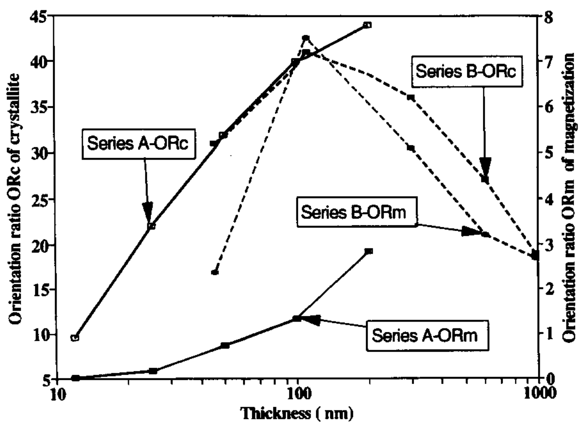

Fig. 3. The orientation ratio OR_c of the crystallites and the orientation ratio $OR_{m\perp}$ of the magnetization as a function of the thickness for Series A and B films.

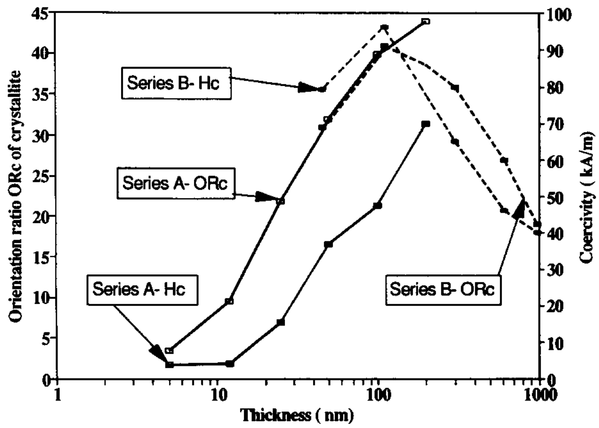

Fig. 4. The orientation ratio OR_c of the crystallites and H_c as a function of the thickness for Series A and B films.

In the range of thicknesses less than 12 nm, a different variation is observed. As the film thickness increases from 5 to 12 nm, the $OR_{m\perp}$ value decreases from 12.5 to 0.18, but the OR_c value increases from 3.5 to 9.6, i.e. the changes of OR_c with the thickness are opposite to the changes of $OR_{m\perp}$ with the thickness. This fact again proves that the thinnest film exhibits the anomalous behaviour.

As pointed out previously, for most Co–Cr films the magnetic properties are mainly controlled by the magnetocrystalline anisotropy. This point of view is also supported by the measured $H_{c\perp}$ and the orientation ratio OR_c of the crystallite as a function of the film thickness, which are shown in Fig. 4. It was found from Fig. 4 that for all $Co_{81}Cr_{19}$ films the variation of $H_{c\perp}$ with the thickness basically follows that of OR_c with thickness, except for films having thickness less than 12 nm. For example, for Series B films, in the range of 46–110 nm, the coercivity $H_{c\perp}$ increases with increase of the orientation ratio OR_c and in the range of 110–980 nm decreases with decrease of the OR_c value. The maximum values of both $H_{c\perp}$ and OR_c appear to be in the of film thicknesses around 110 nm.

On the other hand, for Series A films in the range of 12–200 nm, both $H_{c\perp}$ and OR_c increase monotonically with increasing thickness. However, for $Co_{81}Cr_{19}$ films having thickness less than 12 nm, $H_{c\perp}$ is almost independent of the thickness, even though OR_c still increases with increasing thickness. In addition, the fact that Series B films have larger $H_{c\perp}$ values than those of Series A films again proves that the influence of the substrate on the magnetic property can not be neglected.

The parameter $\Delta\alpha_{50}$ derived from the distribution curve of the crystallite orientation can be considered as another measure of the grain orientation along the in-plane direction. The squareness ratio $R_{s\parallel}$ is thought to be a measure of the in-plane magnetic anisotropy for all grain orientations. In principle, the larger the $\Delta\alpha_{50}$ value, the larger the $R_{s\parallel}$ value will be. The relation between the changes of the measured $R_{s\parallel}$ and the $\Delta\alpha_{50}$ values with the thickness for Series A and B films was plotted in Fig. 5. The curves presented here can be divided into three different types:

(a) For Series A films, in the thickness range of 12–50 nm, both $\Delta\alpha_{50}$ and $R_{s\parallel}$ values decrease with increasing thickness, and there appears to be a good correlation between the $\Delta\alpha_{50}$ and $R_{s\parallel}$ values.

(b) For Series A films which are thinner than 12 nm, as the thickness increases from 5 to 12 nm, the $\Delta\alpha_{50}$ value decreases, whereas the $R_{s\parallel}$ value increases. A similar phenomenon is also observed in Series A films having thickness range of 50–200 nm. This means that on this condition the orientation of the magnetization does not depend on the $\Delta\alpha_{50}$.

(c) For Series B films, as the thickness increases from 110 to 980 nm, the $\Delta\alpha_{50}$ value increases from 3° to 4.5°, but the $R_{s\parallel}$ value is almost constant (= 0.03), i.e. independent of the thickness. Perhaps, this fact implies that in this case, the orientation of the magnetization along in-plane direction in $Co_{81}Cr_{19}$ films is not only determined by the orientation of the crystallite, but also is controlled by other factors, which may include magnetostatic interaction between the neighbouring columnar grains.

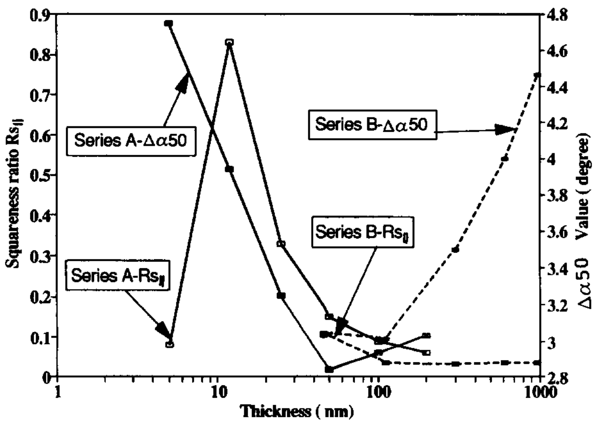

Fig. 5. The measured $R_{s\parallel}$ and $\Delta\alpha_{50}$ values as a function of the thickness for Series A and B films.

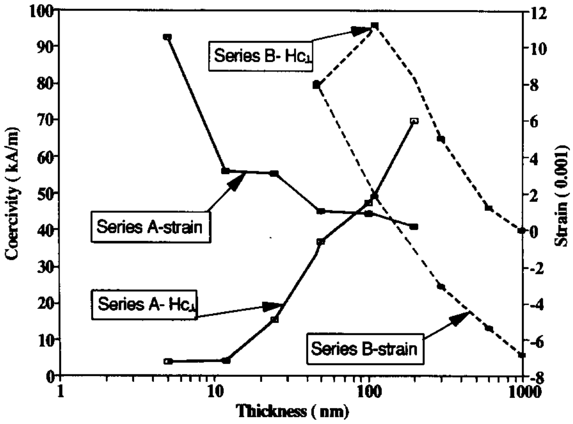

Fig. 6. The relation between the dependence of the measured $H_{c\perp}$ and the film strain ϵ on the thickness for Series A and B films.

The correlation between the measured $H_{c\perp}$ and the strain ϵ with the thickness was illustrated in Fig. 6. As was pointed out [9], for Series A films, as the thickness increases from 12 to 200 nm, the magnitude of the tensile strain decreases monotonously with increasing thickness. Considering that Co–Cr film has a negative magnetostriction constant, this reduction in the tensile strain is favourable for enhancing $H_{c\perp}$. Fig. 6 displays that $H_{c\perp}$ increases with the decrease of the tensile strain. In this case, it seems there is a good relationship between the strain and $H_{c\perp}$. However, in the range of 5–12 nm, the $H_{c\perp}$ for Series A films is independent of the strain. On the other hand, for Series B films if the thickness increases from 46 to 980 nm, the strain in the film gradually changes from the tensile to the compressive strain and in the vicinity of 130 nm the film is in the stress-free state. For example, as the film thickness increases from 130 to 980 nm, the strain increases from zero to -6.8×10^{-3}. In this case, if the magnetostriction anisotropy exerts a dominant influence on $H_{c\perp}$, $H_{c\perp}$ should increase with increasing thickness.

However, experimental results display the opposite tendency. In fact, it is found from Fig. 6 that the maximum $H_{c\perp}$ values for Series A and B films appear in such a thickness range, where the strain ϵ is almost equal to zero.

Considering the fact that the variation of $H_{c\perp}$ with the thickness basically follows the dependence of the orientation ratio OR_c of the crystallite on thickness, instead of the dependence of the strain on the thickness it can be again concluded that for $Co_{81}Cr_{19}$ films thicker than 12 nm, the magnetocrystalline anisotropy makes a decisive contribution to the coercivity.

3.3. The comparison of the calculated and measured texture dependent magnetic properties

In order to calculate the influence of texture on the magnetic properties, a distribution of the crystallographic C-axis orientation $I(\alpha)$ were measured [9]. According to the formula (4), the texture expansion coefficients T_L can be calculated. $F(\alpha)$ is the normalized orientation distribution of (0002) planes, which is derived from the measured $I(\alpha)$. Obviously, if we want to calculate the influence of the texture on $H_{c\perp}$, the effective anisotropy field must be calculated first. To assess the contribution of the magnetocrystalline anisotropy field ($H_a = 2K/M_s$), the shape anisotropy field ($H_d = (N_0 - N_t)M_s$) and the stress anisotropy field ($H_\sigma = 3\sigma\lambda_s/M_s$) to the effective anisotropy field, the shape anisotropy and stress anisotropy will be discussed in some details.

3.3.1. The shape anisotropy

Based on the experimental data [15,16], it was known that the columnar grains in Co–Cr film are surrounded by the Cr-rich boundaries of paramagnetic phase. It is reasonable to assume that these columnar grains in $Co_{81}Cr_{19}$ film form an assembly of non-interacting single domain columns, whose magnetization reversal mechanism is basically controlled by coherent rotation. For simplicity, the shape of the columnar grain in our calculation is considered to be a cylinder. For a single-domain column with the cylindrical shape, the demagnetizing factor along the length is a function of the aspect ratio (the ratio of the length to the diameter of the columnar grain), and can be found in references [17,18]. A typical values of the demagnetizing factor are listed in Table 3.

Table 3
The demagnetizing factor for a cylindrical particle

Aspect ratio	N_0 (10^{-2})
1	24
1.5	15
2	10.6
3	5.9
4	3.7
5	2.5
6	1.8
8	1.06
10	0.69
15	0.32
20	0.17
30	0.043

According to the experimental data of the correlation between the aspect ratio and the thickness [9] as well as the first order magnetocrystalline anisotropy constant K_1, the effective anisotropy field ($2T/M = H_a + H_d$) can be calculated, if the magnetostriction effect is neglected. Then, according to the formula (2) the $H_{c\perp}$ as a function of the thickness is calculated.

For Series A and B films, the calculated $H_{c\perp}$ values for different thicknesses calculated without taking into consideration of the magnetostatic interaction between the columns were plotted in Fig. 7. In order to compare the calculated $H_{c\perp}$ with the measured $H_{c\perp}$ values, the measured $H_{c\perp}$ as a function of the thickness was drawn in Fig. 7. It can be seen from Fig. 7 that the changes of the calculated $H_{c\perp}$ with the thickness are in a good agreement with experimentally observed changes. However, there are a large difference in

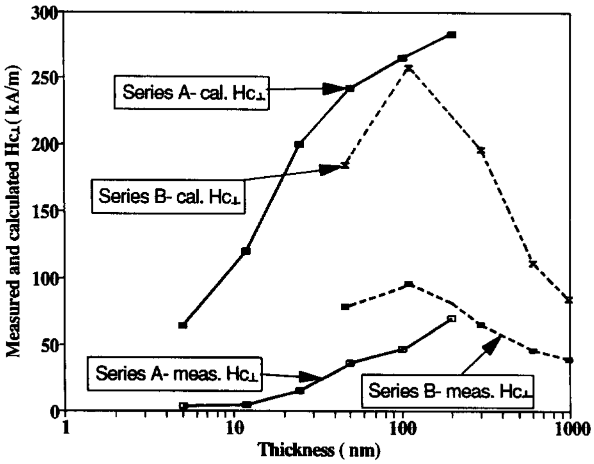

Fig. 7. The measured $H_{c\perp}$ and the calculated $H_{c\perp}$ values (without magnetostatic interaction between columns) versus thickness for Series A and B films.

magnitude between the measured and the calculated $H_{c\perp}$ values. This fact might indicate that the influence of the magnetostatic interaction on the calculated $H_{c\perp}$ should not be neglected and the magnetization reversal tends to the incoherent rotation mode.

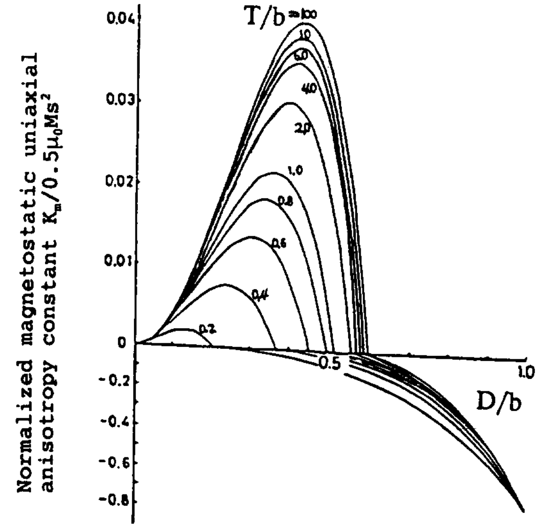

Fig. 8. The magnetostatic uniaxial anisotropy constant K_m normalized by demagnetizing energy $0.5\mu_0 M_s^2$ (after [19]). D: the diameter of the cylinder; T: the length of the cylinder; b: the spacing between neighbouring cylinders.

Method of calculations of the uniaxial anisotropy resulting from the magnetostatic energy in magnetic film with a columnar structure was in-

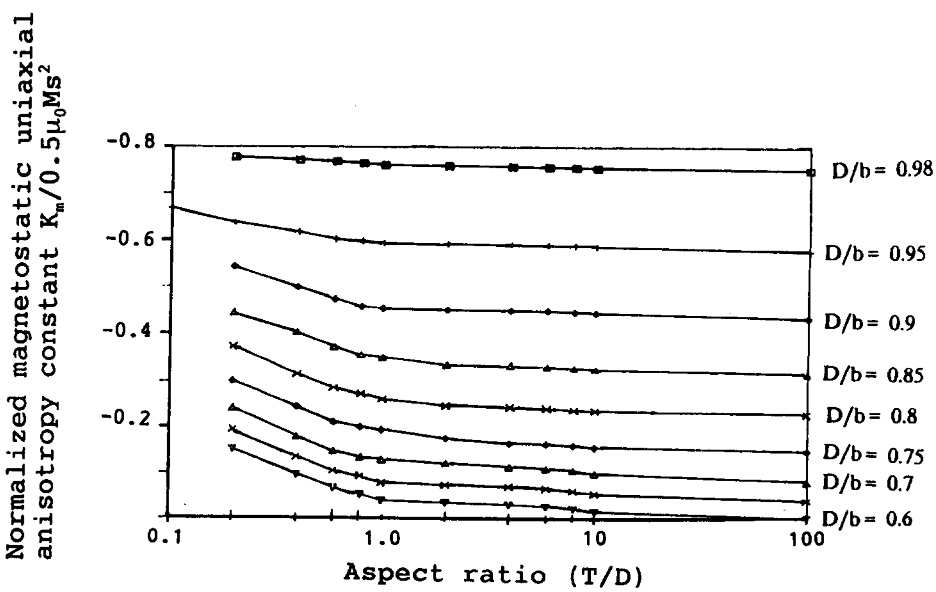

Fig. 9. The K_m value as a function of the aspect ratio for different film densities (D/b).

troduced in reference [19]. This has been worked out in more detail in [20]. In proposed models small magnetic cylinders were assumed to form an infinite two-dimensional array having hexagonal symmetry. Based on the formula used for calculating the uniaxial anisotropy constant K_m the dependence of the $K_m/0.5\mu_0 M_s^2$ on the ratio D/b [19] (the 'packing density' of magnetic particle) is shown in Fig. 8. The ratio T/b is a parameter of curves, where T and D are the length and diameter of the cylinders respectively, and b is the spacing between neighbouring cylinders. As shown in this figure, the uniaxial easy direction of the magnetization might be normal to the film plane or may lie in it depending on whether the K_m is positive or negative. In general, for a low 'packing density' film the orientation of the magnetization aligns along the film normal. On the contrary, for a high 'packing density' film the magnetization is oriented along the in-plane direction. There is a transition point for each particular ratio of T/b.

Considering that in Co–Cr films the 'packing density' is usually close to 1, for a film having high density the relationship between K_m and the aspect ratio (T/D) can be shown in Fig. 9. It can be seen from Fig. 9 that the larger the film density the larger is the negative magnitude of the magnetostatic uniaxial anisotropy constant. The aspect ratio T/D has only a small influence on the K_m value as compared to the influence of the film density.

Based on the statistical results of TEM observations [16,20], it was estimated that the 'packing density' of Co–Cr films is in the range of 0.85–0.96, if the grain boundary is assumed to be the non-magnetic area. In order to fit well the calculated $H_{c\perp}$ value to the measured one, using our calculation procedure the film densities are assumed as 0.98 and 0.95 for Series A and B films. According to the formulae (2), (3) and (4) as well as under the prerequisite that $H_d = 2K_m/M_s$, the $H_{c\perp}$ values for Series A and B films were calculated. The changes of the measured and calculated $H_{c\perp}$ with thickness were plotted in Fig. 10. Comparing the Fig. 7 with the Fig. 10, it can be clearly seen that the $H_{c\perp}$ values calculated using the assumption of the influence of the magnetostatic interaction are much closer to the measured $H_{c\perp}$ values than those obtained when neglecting the interaction effect, exception are for films thinner than 12 nm. However, as we have mentioned previously, the thinnest films exhibit anomalous behaviours.

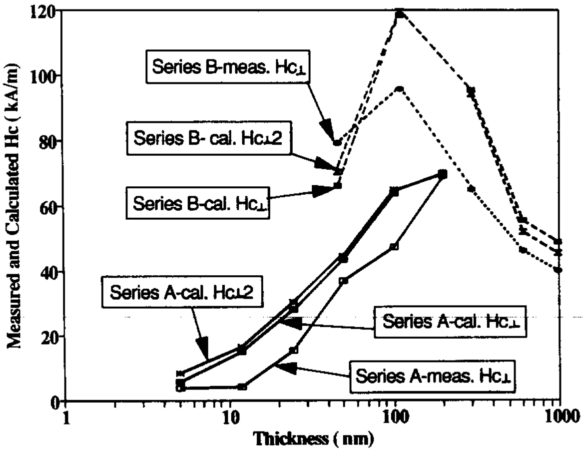

Fig. 10. The measured $H_{c\perp}$, the calculated $H_{c\perp}$ (with magnetostatic interaction) and the calculated $H_{c\perp 2}$ (with interaction and magnetostriction effects) for Series A and B films.

3.3.2. The magnetostriction anisotropy

Based on the measured changes of the film strain ϵ as a function of the thickness, for Series A and B films the magnetostriction anisotropy constant K_σ could be calculated, assuming that the Young's modulus is 21×10^{11} dyn/cm^2 and the saturation magnetostriction constant is -12×10^{-6} [6]. The calculated magnetostriction constant K_σ is shown as a function of the thickness in Fig. 11. For Series B films, as the film thickness increases from 46 to 980 nm, the film magnetostriction constant K_σ gradually changes from negative to positive value and in the vicinity of 130 nm becomes zero. This fact suggests that the existing magnetostriction anisotropy can strengthen perpendicular anisotropy in thick films. On the contrary, the magnetostriction anisotropy in thinner films (< 130 nm) will weaken the perpendicular magnetic anisotropy. For Series A films, the magnitude of negative magnetostriction constant decreases monotonically with increasing thickness. Based on the

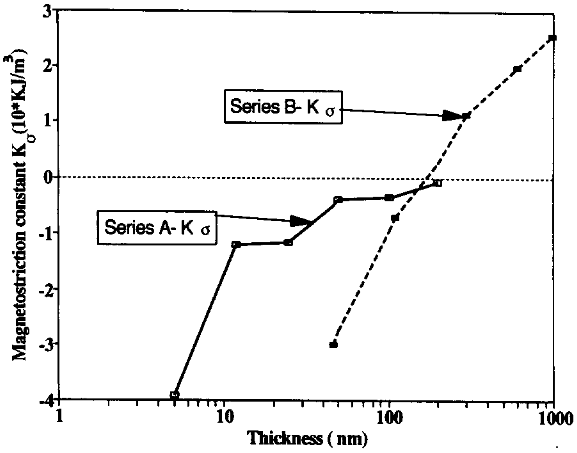

Fig. 11. The magnetostriction constant K_σ for Series A and B films.

above results, if the magnetostriction constant K_σ is used to calculate the magnetostriction anisotropy field H_σ, the $H_{c\perp}$ value can be calculated. However, considering the fact that the maximum $H_{c\perp}$ value for both Series A and B films appears in such a thickness range, where the K_σ values are close to zero and in order to fit the calculated $H_{c\perp}$ value to the measured value, it was found that in calculating the effective anisotropy field ($2T/M = H_a + H_d + gH_\sigma$) the fitting coefficient $g = -0.1$ should be used, the reason for which still remains to be investigated. The changes of $H_{c\perp}$ with the thickness calculated assuming both the effects of the magnetostriction anisotropy and the magnetostatic interaction were compared to experimental results in Fig. 10. From this figure the conclusion can be made that the magnetostriction anisotropy has only a minor influence on the coercivity $H_{c\perp}$ compared to the magnetocrystalline and shape anisotropy. The comparison between the calculated and measured results of H_c vs. thickness implies that the presence of magnetostatic interactions is favourable for forming the incoherent rotation.

Finally, let us review again the dependence of $H_{c\perp}$ on the thickness for $Co_{81}Cr_{19}$ films. According to the above-mentioned results, the curves presenting the variation of $H_{c\perp}$ and $R_{s\parallel}$ with thickness are plotted in Fig. 12. So far as the $H_{c\perp}$ vs. thickness curve is concerned, a different behaviour was observed into two different areas:

(a) In the thickness range of B–E, the $H_{c\perp}$ vs. thickness curve is similar to that of the orientation ratio OR_c vs. thickness curve. There is a critical thickness, at which maximum $H_{c\perp}$ and the maximum OR_c were observed. In this thickness range, the magnetocrystalline anisotropy plays a dominant role.

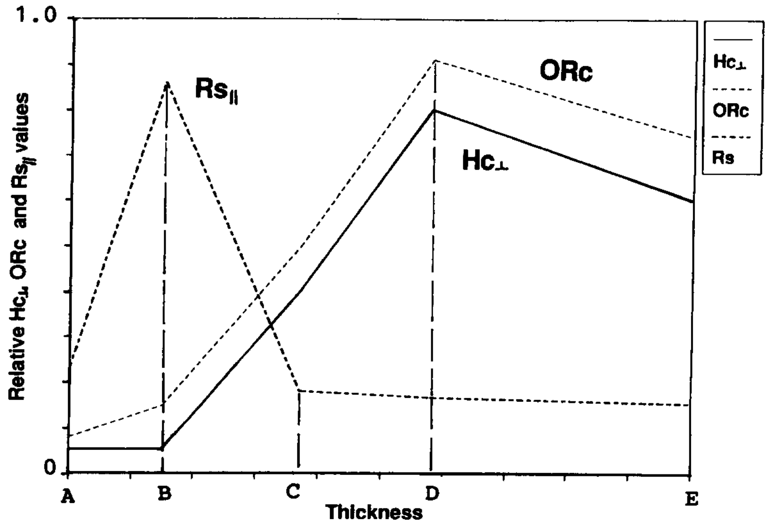

Fig. 12. The schematic presentation for $H_{c\perp}$, $R_{s\parallel}$ and OR_c vs. thickness for $Co_{81}Cr_{19}$ films.

(b) In the thickness range of A–B, the $Co_{81}Cr_{19}$ films exhibit an anomalous behaviour. The following observations can be made:

(1) The $H_{c\perp}$ is almost independent of the thickness.

(2) The $R_{s\parallel}(M_r \parallel / M_s)$ value increases with increasing thickness, although the orientation ratio OR_c increases with increasing thickness.

4. Conclusions

(1) The shape of the hysteresis loop for $Co_{81}Cr_{19}$ films depends on the film thickness. In the thickness range of 5–12 nm as the thickness increases, the shape of the hysteresis loop gradually changes from a square loop in the perpendicular direction to an in-plane square loop, indicating that the $Co_{81}Cr_{19}$ film changes from a perpendicular to a longitudinal anisotropy medium. For $Co_{81}Cr_{19}$ films thicker than 12 nm, as the thickness increases, the films again change from a longitudinal anisotropy medium to a perpendicular anisotropy one. In general, their hysteresis loops exhibit a good rectangular parallelepiped shape.

(2) For $Co_{81}Cr_{19}$ hcp films the textural evolution exerts a tremendous influence on the magnetic properties. In the thickness range of (12–982 nm) the changes of both $H_{c\perp}$ and the magnetization orientation ratio $OR_{m\perp}$ with the thickness follow the same change in the pattern of the orientation ratio OR_c of the crystallite very well, indicating that in this case the orientation of the texture plays a dominant role in the determination of the magnetic properties. It was found that for Series B films the maximum of $H_{c\perp}$, $OR_{m\perp}$ and OR_c values appear in the vicinity of 130 nm. This means that a high quality medium for perpendicular recording can be made by depositing a $Co_{81}Cr_{19}$ film having such a thickness. However, in the thickness range of 5–12 nm the $Co_{81}Cr_{19}$ films exhibit an anomalous behaviour. The anomalous behaviour is expressed by the following observations: (a) as the thickness increases, $H_{c\perp}$ is almost independent of the thickness; (b) the squareness ratio $R_{s\parallel}$ increases with increasing thickness, even though the orientation ratio OR_c of the crystallite still increases with increasing thickness.

(3) For Series B films as the thickness increases, the film magnetostriction anisotropy constant gradually changes its value from negative to positive. For Series A films the magnitude of negative magnetostriction constant decreases monotonically with increasing thickness. In the range of 110–150 nm, the magnetostriction constants for Series A and B films are close to zero.

(4) Based on the textural evolution observed as the thickness changes, for films thicker than 12 nm, calculations of the coercive force were made under the assumption that the magnetization process is dominated by rotational reversal and that the columnar grains are single-domain fine particles oriented towards the external field. Experimental data show that the calculated variation of $H_{c\perp}$ with thickness are in much better agreement with the experimental results, if magnetostatic interaction is taken into consideration. This suggests that in this case the coercivity $H_{c\perp}$ is mainly controlled by both magnetocrystalline and shape anisotropy and the magnetization reversal behaviour is considered as incoherent.

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