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Effect of grain misorientation on the stripe domains in evaporated cobalt films

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Abstract. X-ray diffraction and magnetic force microscopy techniques were used to investigate the structural and the static magnetic properties of vapor-deposited cobalt films with various thicknesses t ranging from 50 to 195 nm. Texture measurements revealed that as the thickness increases, the films become predominantly c-axis oriented. Magnetic stripe domains structure was only observed for the thicker films, with $t = 195$, 173 and 125 nm, while such a magnetic configuration was expected for all the samples based on the theoretical studies. Since the layers present increasing c-axis misorientation when the thickness decreases, we assume that this effect can prevent the stripe domains formation. This behavior is qualitatively explained by a simple model which describes the stripe domains structure taking into account the role of a small misorientation of the anisotropy axis.

PACS. 61.10.Nz X-ray diffraction – 68.37.Rt Magnetic force microscopy (MFM) – 75.70.Kw Domain structure (including magnetic bubbles)

1 Introduction

Cobalt and magnetic systems containing cobalt (Co) have been extensively studied both fundamentally and practically. At present, they are the dominant materials used for longitudinal magnetic recording [1]. In hcp cobalt films, both the c-axis orientation and the grain structure contribute to the presence of a magnetocrystalline anisotropy [2]. It is one of the important ingredients responsible for the development of real stable magnetic domains in ferromagnetic materials, the nature of wich depends on the film thickness and on the magnetic history [3]. In a properly demagnetized sample, the domains structure corresponds to the minimum of its total energy.

The periodic stripe domains structure is characteristic of ferromagnetic films with a perpendicular anisotropy field H_a [3–5]. Among the various types of stripe domains, one can mention the so-called weak stripe domains appearing in magnetic films with a low or intermediate perpendicular anisotropy ($Q < 1$, with Q the quality factor defined as $Q = H_a/4\pi M$, where M is the saturation magnetization). A second class of stripe domains [6,7], wich are not formed spontaneously as regular stripes, develop rather by nucleation and growth. This type of stripe domains has been often called strong stripe domains because of their contrast in the electron microscope.

When a magnetic layer is subjected to a sufficiently strong in-plane external field **H**, the equilibrium state is associated to a uniform magnetization parallel to **H**. For lower fields, the equilibrium state in a layer with a perpendicular anisotropy can be modified into stripe domains configuration corresponding to a periodic out of plane magnetization precessing around the direction of the applied field [3,8]. This physical process is a way to save a part of the anisotropy energy. For this kind of equilibrium state, the required conditions were first discussed by Muller [9]. In this approach, the minimal thickness of the layer for which small magnetization oscillations are possible is computed. This minimal thickness t_c defines a limit below which no stripe domains can be formed. The computation method uses the linearized Landau-Lifshitz equations of motion for a layer defined by given magnetic parameters (exchange stiffness A, perpendicular magnetocrystalline anisotropy K_u(with $H_a = 2K_u/M$), saturation magnetization M) in the presence of an external field H.

The film deposition conditions which monitor the formation and the nature of the texture also influence this limit. For instance, values for t_c of 40 nm [4] or 25 nm [5] have been derived from the investigation of magnetic domains of epitaxially hcp Co films grown by molecular beam epitaxy (MBE) technique. In vapor-deposited films, the formation of fiber texture is governed by the evolutionary selection model proposed in the late 1960's by

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Van der Drift [10]. It implies that out of plane texture (fiber) is stronger as the metallic film thickness increases. The vapor-deposited Co films under investigation in this study follow such a behavior as it will be shown later. They all meet the $Q < 1$ condition as deduced from their determined magnetic parameters [11]. According to Muller's criterion, they all should exhibit a stripe domains structure at equilibrium. However, we did observe such magnetic structure only for the thicker films.

The aim of this paper is to propose a qualitative understanding of this behavior. We present X-ray diffractometry (XRD) texture measurements and magnetic force microscopy (MFM) investigations of the Co films with thickness t ranging from 50 to 195 nm. The relationship between the observation of static equilibrium magnetic structures and the film textures are discussed within the framework of a simplified description of the stripe domains taking into account the role of a small misorientation of the anisotropy axis.

2 Experimental details

The cobalt samples under investigation were obtained by evaporation onto a silicon (100) substrate at room temperature with base pressure during deposition lower than 10^{-6} mbar. The deposition rate was of few Å/s and the films thicknesses were controlled from profilometer depth measurements. XRD scans ($\theta/2\theta$) and texture measurements were achieved using a four-circle goniometer (INEL) equipped with a Co K_α source in the Bragg-Brentano type geometry.

The study of the zero-field magnetic structure of the Co films was carried out by MFM measurements using a Veeco 3100 apparatus. We used CoCr-coated cantilevers supplied by Digital, with the tips magnetized along their axis (perpendicular to the sample surface). Before the observations, the films were demagnetized by a slowly decreasing ac field parallel to the sample, which is commonly considered to yield the lowest total energy domain structure.

3 Results and discussion

The studied Co films are polycrystalline which exhibit a hexagonal phase mainly due to the presence of the diffraction peaks ⟨10-10⟩, ⟨0002⟩ and ⟨10-11⟩ respectively at values of θ equal to 24.4°, 26.2° and 27.9° (i.e. 2θ equal to 48.8°, 52.5° and 55.8°). These peaks denoted 1, 2 and 3 in the X-ray spectra, are shown in Figure 1 (a, b, c). This figure reveals the significant reduction of the intensity of the dominant ⟨0002⟩ peak of the cobalt hexagonal phase for the 125 nm-thick sample. This reduction is due to the combined effect of the thickness decrease and of the microstructure evolution, as discussed in the next paragraph. XRD $\omega/2\theta$ spectra, obtained in this study by using another setup utilizing a grazing incident Cu K_α source, confirmed these structural properties but did not

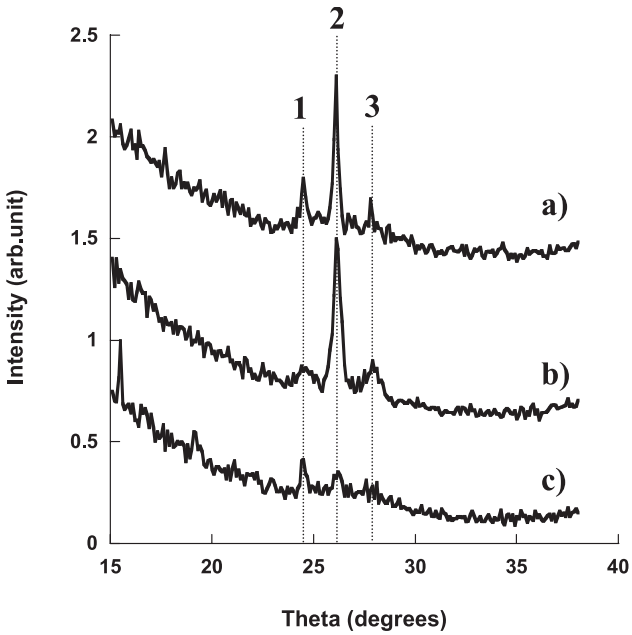

Fig. 1. XRD scans for the sample with thickness $t = 195$ nm: a), $t = 173$ nm: b) and $t = 125$ nm: c). The diffraction peaks assigned as 1, 2 and 3 in the X-ray spectra correspond to the ⟨10-10⟩, ⟨0002⟩ and ⟨10-11⟩ Co peaks.

provide additional information in the case of the 50 and 70 nm-thick films. These thinnest films showed a similar experimental behavior but, due to the weakness of the intensity, the data were less exploitable for texture measurements. In the following, we focus on the three thickest films (i.e. 195, 173 and 125 nm-thick samples). For these samples, the signal to noise ratio was satisfactory enough to draw clear and unambiguous information about the texture of the films. In this analysis, each sample is associated, by convention, with a macroscopic orthonormal reference mark formed by three directions, entitled, respectively, normal, rolling and transverse directions (i.e. ND, RD, TD). The analyzed face is the one perpendicular to ND. For the thickest layers (125 to 195 nm), the acquisition of three pole figures (with a maximum tilt angle of 80°) respectively on the plane families {10-10}, {0002}, {10-11} of the hexagonal phase enabled us to determine in each case the orientation distribution functions (ODF). Based on the ODF functions, the inverse pole figures are then obtained in the normal direction using the vector method [12]. Pole figures and inverse pole figures are two dimensional projections of the three dimensional ODF.

The analysis of the inverse pole figures shows the presence of a fiber texture (no preferential orientation in the RD-TD planes with a preferential orientation of the ⟨0002⟩ directions parallel to ND direction, with a misorientation of about 10° (Figs. 2 a, b, c). This means that the films are predominantly c-axis oriented. We note that the intensity of the maximum, always located in the fiber, decreases with the film thickness as follows: (i) 9.9 for the 195 nm-thick sample; (ii) 8.0 for the 173 nm one; and (iii) 7.7 for the 125 nm-thick one. We also observe that the fiber

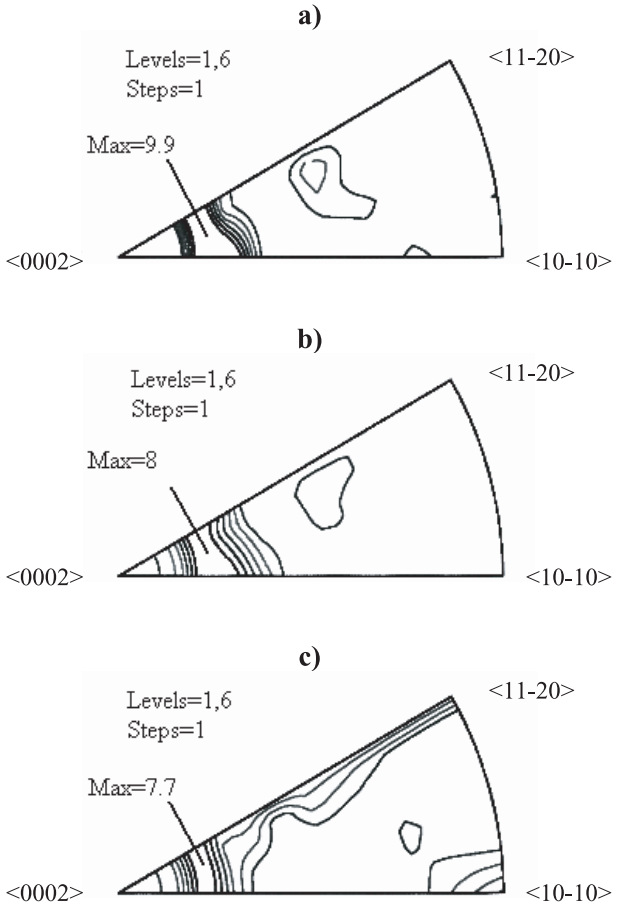

Fig. 2. Inverse pole figures for the sample with thickness $t =$ 195 nm: a), $t = 173$ nm: b) and $t = 125$ nm: c). The intensity of the maximum is mentioned for each sample.

broadening, respectively, 10°, 15° and 25° for the 195, 173 and 125 nm-thick films increases when the film thickness decreases. This behavior is in agreement with the Van der Drift evolutionary model which stipulates that texture is stronger when the metallic film thickness increases [10]. We can also observe in Figure 2c that, in the case of the 125 nm sample, the ⟨0002⟩ texture component decreases to the benefit of the ⟨10-10⟩ component and other random orientations of the grains. This behavior, combined with the reduction of the film thickness, contributes to the decreasing intensity of the ⟨0002⟩ X-ray peak observed in Figure 1c.

We consider now the MFM results. The zero-field equilibrium configuration for the magnetization of the Co layers has been studied after their demagnetization by a slowly decreasing 50 Hz ac field. This method has been suggested as being one of the best in order to obtain the lowest possible energy domain configuration [13].

Experimentally, we only observed the stripe domains configurations for the thicker films with a thickness of 195, 173 and 125 nm respectively. The corresponding MFM images are shown in Figure 3. The exhibited configuration is characteristic for a nonzero perpendicular magnetization component oriented alternatively up and

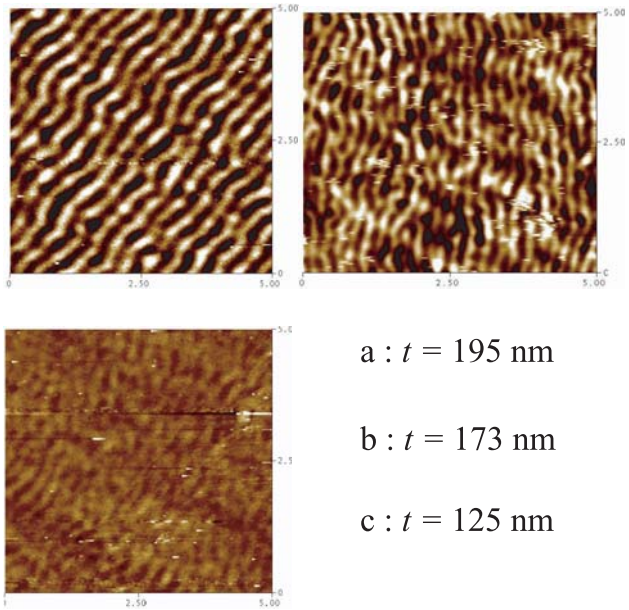

Fig. 3. Zero-field MFM images 5 μm × 5 μm obtained after parallel ac demagnetization.

down. These films present a positive perpendicular magnetic anisotropie K_u, larger than 10^6 erg/cm^3 [11]; the observation of the magnetic stripe domains was then expected. It has been previously shown in Co-Ag granular thin films [14,15] that the MFM observation of stripe domains patterns is strongly linked to the sign and to the magnitude of the perpendicular magnetic anisotropy.

From our MFM images, the stripe width is found to be comparable in size to the film thickness as theoretically predicted for low Q systems [3]. This condition is satisfied by our films [11]. The MFM contrast for the 125-nm-thick film is weaker, despite a significant stronger applied ac demagnetizing field. In fact, to observe stripe domains, we had to increase the initial value of the decreasing ac field (amplitude 1000 Oe compared to 100 and 200 Oe for the 195 and 173 nm-thick films respectively). The 70 and 50-nm-thick samples require an ac field stronger than what is technically available in our experimental facilities. After the strongest ac demagnetization (amplitude 2000 Oe), we observed a magnetic structure which seems to be a compromise between the saturation state and a stripe domains configuration as shown in Figure 4 for the 50 nm-thick film.

In order to forecast whether stripe domains should be observed in our samples, we used the Muller's criterion [9]. It allows to draw a diagram representing the evolution of the reduced parameter t_c/t_m as a function of the quality factor Q, where t_c is the critical thickness and t_m is defined as $2\pi(A/K_u)^{1/2}$. Figure 5 shows this evolution for a zero external applied field. The continuous line defines two areas: (i) one area where stripe domains can appear (upper area); and (ii) another area where they cannot be formed (lower area). According to the previously measured values of the magnetic parameters M, K_u and A [11], all our samples should spontaneously exhibit such

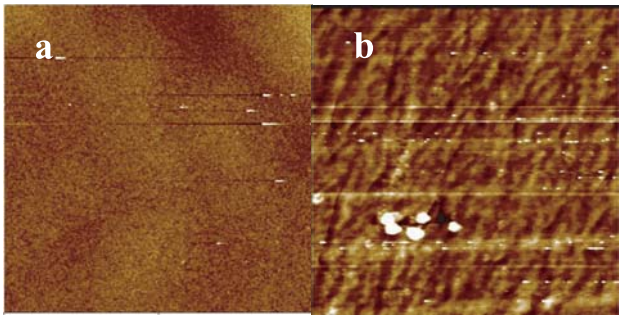

Fig. 4. Zero-field MFM images obtained for the 50 nm-thick layer [a: 5 μm × 5 μm (remanent state) ; b: 10 μm × 10 μm (parallel ac demagnetization)].

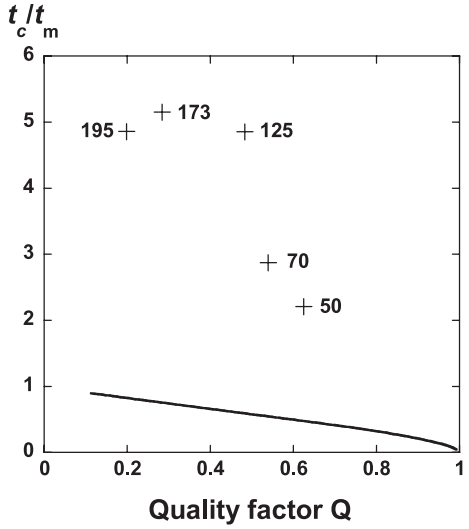

Fig. 5. Graphic showing the two areas bounded by the critical thickness (continuous line). In the upper area, weak stripe domains can be observed. The + symbols represent the positions of our samples as deduced from their characteristic magnetic parameters.

structures, as depicted in Figure 5. This is clearly not the case since we only observed them for the three thickest samples. Since the X-ray texture analysis shows that the samples present an increasing c-axis misorientation when the thickness decreases, we can assume that the stripe domains formation is prevented by such an effect. For this purpose, we propose a simplified description of the stripe domains which includes the role of a small misorientation of the anisotropy axis.

In stripe domains configuration, beyond a critical thickness, the magnetization is almost parallel to a direction in the plane of the layer. However, as mentioned in the introduction, there also exists a small oscillatory component m perpendicular to the main component M. This small component is alternatively perpendicular or parallel to the layer. We can assume that m makes Landau domains as shown in Figure 6. This corresponds to the simplest way to describe the stripe domains [3]. We also assume that m is zero in the cores around which it turns. Finally, we suppose that the layer is characterized by a set

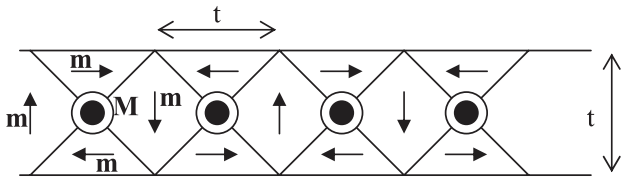

Fig. 6. Approximative configuration of weak stripe domains: Landau domains as induced by the small oscillatory component m perpendicular to the main component M with the assumption that m is null in the cores around which it turns.

of magnetic parameters, namely M, K_u and A. For a very small value of m, we can write the energy of the volume $V = t{\times}\mathrm{t}{\times}L$ shown in Figure 6 as follows:

$$E = -(m/M)^2 K_u L t^2/2 + (m/M)^2 AL \qquad (1)$$

where t is the thickness (which is also the half period) and L is the length of the stripe. The first term in equation (1) is the energy gained by the part where the small component is perpendicular to the layer (i.e. parallel to the anisotropy axis). The second term is the energy provided to make the magnetization turn by an angle (m/M) from the core where $m = 0$ to the neighboring regions where it is not null. This term should be multiplied by a numerical factor because a core does not exactly cost $(m/M)^2$ AL but we assume that the parameter A includes the right numerical factor. Finally, equation (1) becomes

$$E = (m/M)^2(2A - K_u t^2)L/2. \qquad (2)$$

It is evident from equation (2) that m is spontaneously not null if $t^2 > 2A/K_u$. This defines a critical thickness beyond which the stripe domains exist.

Now we assume that the anisotropy axis is locally tilted by an angle a (Fig. 7a). Therefore, the gained energy per unit volume in a region where m is perpendicular to the layer is

$$K_u((m - aM)/M)^2 = K_u((m/M)^2 - 2a(m/M) + a^2). \quad (3)$$

Consequently, we assume that the effect of the variation of the direction of the anisotropy axis can be taken into account by writing the mean energy of a volume $V = t \times t \times L$ as follows:

$$E = E^\circ - ((m/M)^2 - 2\underline{a}(m/M))K_u L t^2/2 + (m/M)^2 AL. \qquad (4)$$

Which can be rewritten as

$$E - E^\circ = (m/M)^2(2A - K_u t^2)L/2 + (m/M)K_u t^2 L\underline{a}. \quad (5)$$

The variation of $E - E^\circ$ is displayed in Figure 7b for $t^2 > 2A/K_u$. The energy is minimized for large value of (m/M) but the energy $E_{max} - E^\circ$, given by equation (6), must be first provided.

$$E_{max} - E^\circ = L(K_u t^2 \underline{a})^2/(2K_u t^2 - 4A). \qquad (6)$$

It is is clear from equation (6) that the required energy increases with the misorientation $\underline{a}$. This simple model

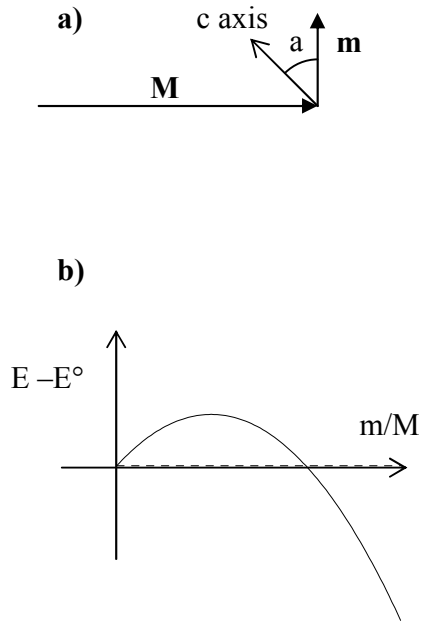

Fig. 7. a) Representation of the local tilt (angle a) of the anisotropy axis modifying the anisotropy energy from $-K_u(m/M)^2$ to $-K_u((m-aM)/M)^2$ (see text), b) $E - E^\circ$ variation in the case $t^2 > 2A/K_u$. The energy is minimized for large value of (m/M) but the energy $E_{max} - E^\circ$ must be first provided.

can explain the reasons for being able to observe stripe domains only for the thick samples corresponding to a small c-axis misorientation. This behavior is supported experimentally by the fact that as the sample thickness decreases, the misorientation increases requiring a larger initially applied ac field.

4 Conclusion

We tried in this work to show the influence of the texture on the formation of stripe domains in magnetic films exhibiting a perpendicular anisotropy. A simplified description of the stripe domains including the role of a small misorientation of the anisotropy axis showed that the formation of such a magnetization equilibrium configuration can indeed be prevented by a more or less strong misorientation of the grains.

The structural properties and magnetic history of the films are key parameters which dominate the stripe domains formation. The thickness at which the stripe domains structure first appears is sensitive to the growth conditions and to the crystalline properties of the films.

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