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Influence of thermal annealing on the morphology and magnetic domain structure of Co thin films

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Abstract

Centimeter scale cobalt films with various thicknesses (8 nm ~ 100 nm) were deposited by electron beam evaporation (EBE) and then annealed in a gas mixture of Ar and H2 at temperatures ranging from 200 °C to 500 °C. Advanced characterization techniques (e.g., XRD, SEM, AFM and MFM) were employed to investigate the influence of annealing on the morphology, crystal structures and magnetic domain structures of Co thin films. The results of SEM and AFM suggest that there is no obvious change in the morphology of Co film before and after annealing especially for thicker films and the root–mean–square roughness of Co film surface is slightly reduced after annealing. The influence of thermal annealing on the magnetic domain structure of EBE Co thin films was investigated by magnetic force microscopy (MFM) for the first time. It is found that even if there is no clear domain structure in the as-deposited films, it is possible to obtain periodic stripe domains with perpendicular magnetic anisotropy (PMA) by thermal annealing owing to the development of HCP Co phase, which was confirmed by XRD analysis. The correlations between the film morphology, thickness and magnetic domain structure are discussed qualitatively. Based on the periodic stripe magnetic domains, the domain–wall energy density of annealed cobalt films is calculated and this study found that the annealing has almost no effect on the energy density of the domain–wall. This work provides an effective way to obtain the perpendicular magnetic anisotropy (PMA) for practical applications.

1. Introduction

Cobalt is one of the most commonly used magnetic metal with high saturation magnetization (1400 emu cm⁻³ [1]) and high Curie temperature (~1145 °C [2]). Among the previous studies of Co films, the magnetic domain structure has always attracted much interest in the last few decades [3–14], since it is a key element of magnetic microstructure associated with magnetic properties, such as magnetic anisotropy [7] and magnetoresistance [15]. The magnetic domain is also a basic unit to build the racetrack memory [16, 17] and domain wall logic [18]. Distinct from the films with strong perpendicular magnetic anisotropy (PMA) (e.g., Nd₂Fe₁₄B [19] and SmCo₅ [20]), the cobalt film has a not very strong uniaxial anisotropy [21], which makes the Co films may exhibit various domain structures depending on their shape, thickness and crystal structure. For example, with the change of thickness, the cobalt film may present in-plane magnetized domains, perpendicular stripe-like domains or more complex domain structures [6].

Up to now, cobalt films can be prepared by different methods, such as molecular beam epitaxy (MBE) [7], metal organic chemical vapor deposition (MOCVD) [4], magnetron sputtering (MS) [5], pulsed laser deposition (PLD) [21], electrodeposition [12–14], thermal evaporation (TE) [9–11] and electron-beam evaporation (EBE) [3, 6]. The EBE process is a physical-vapor-deposition (PVD) method based on the evaporation of metal which is directly heated by the electron beam. Due to its versatility, wide applicability and controllability, EBE is a popular approach in the field of micro- and nano-fabrication techniques [22].

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Annealing is a useful post-fabricating process to reduce internal defects and residual stress in a material, which may influence the domain structures of magnetic films [23–25]. To our best knowledge, several previous studies have investigated the annealing effect on the crystal structure and magnetic properties (e.g., coercivity and hysteresis loops) of Co thin films [26–28]. Sharma et al. [26] focused on the changes of coercivity and resistivity for sputtered Co thin films after annealing. Kumer et al. [27] investigated the evolution of crystal structure, morphology and coercivity with annealing temperature for sputtered Co films. Jergel et al. [28] reported the variation of crystal structure, residual stress and magnetic anisotropy for evaporated Co films. However, few work has addressed the influence of annealing on the magnetic structures of electron-beam evaporated Co films through a magnetic force microscopy (MFM) measurement.

In this work, we prepared electron-beam evaporated cobalt thin films with various thicknesses from 8 nm to 100 nm. Annealing up to 500 °C was performed in a gas mixture of Ar and H2. Scanning electron microscope (SEM) and atomic force microscope (AFM) were exploited to explore the morphology changes of Co films before and after annealing. The annealing effects on the magnetic domain structure was investigated through a magnetic force microscope (MFM) by studying the change of the domain patterns of as-deposited and annealed Co films. The crystal structures for the Co thin films before and after annealing was analyzed by x-ray diffraction (XRD). At last, the relationship between the morphology, thickness and domain structure of Co films has been studied. Using classical physical models and magnetic domain patterns, the effect of annealing on the domain wall energy of Co films was also studied.

2. Experimental

The cobalt thin films with thickness ranging from 8 nm to 100 nm were deposited on high-resistivity (111) Si or SiO2 substrate in an electron beam evaporation (EBE) system (DE-200, DE Technology Ltd.) at room temperature. The basic vacuum of the deposition chamber was 10−7 Torr and the deposition rate was kept to be 0.5 Å s⁻¹. The deposition rate and final film thicknesses were monitored by a quartz crystal microbalance. Once the cobalt films were taken out from the vacuum deposition chamber, they were rapidly immersed into acetone (analytical pure) to avoid oxidation of Co in air.

The annealing process was performed in a quartz tube furnace (Lindberg Blue M, Thermo Scientific) at a suitable temperature for 1 h in atmospheric pressure using a gas mixture of 3% hydrogen and 97% argon. Co films with different thicknesses on Si substrate were annealed at 300 °C, and other samples with thickness of 50 nm on SiO2 substrate were independently annealed at 200 °C, 400 °C and 500 °C, respectively. Before starting the annealing process, the tube furnace was vacuumed by a mechanical pump and refilled with the gas mixture (3% H2 and 97% Ar) for three times to eliminate air. During the annealing process, the gas mixture was kept flowing at a rate of 200 sccm to provide efficient protection from oxidation.

The morphology of Co films was characterized by means of scanning electron microscopy (SEM, Zeiss Sigma 300) and atomic force microscopy (AFM, Bruker Dimension Icon). For the SEM imaging, accelerating voltage (EHT) of 5 kV and working distance (WD) ranging from 3 mm to 5 mm were applied with an SE2 detector. AFM measurements were accomplished with Tapping mode using Bruker OTESPA probes. The chemical composition of the Co film was verified by energy dispersive spectroscopy (EDS) at an acceleration voltage of 10 kV or 15 kV in the SEM assembling with a SmartEDX. The crystal structure was determined by an x-ray diffractometer (XRD, PHILIPS X’Pert-MRD) with Cu Kα radiation (λ = 1.54 Å). It should be noted that in addition to XRD, another effective method to determine the crystal structure of Co thin films is high resolution transmission electron microscopy (HRTEM), which can be used to further confirm the XRD analysis results. Unfortunately, when preparing TEM samples, we failed to find an effective method to firmly fix the magnetic cobalt films in the TEM holder. In order to avoid the risk that the cobalt film may contaminate or damage the pole-pieces or optical channel in the electron microscope, we have to give up the TEM and use XRD analysis to study the crystal structure changes of the Co film before and after annealing. From previous work, it seems also possible to use XRD alone to determine the crystal structure of the cobalt film, as in [3, 8, 11 and 21, 3, 8, 11, 21].

Magnetic domains of Co thin films were imaged by a magnetic force microscopy (MFM, Bruker Dimension Icon), which is one of the most commonly used tools for the observation of domain structure [29–31]. MFM is based on the LiftMode, the process of which can be described as follows: First, a tapping cantilever equipped with a magnetic tip tracks the surface to record the topography information; Second, the tip is then raised to a certain height (so-called lift height) above the sample surface; Third, the tip profiles the topography obtained in the first scan and simultaneously monitor the influence of magnetic forces. The magnetic force gradient can be detected through an oscillated cantilever in three ways: phase detection, amplitude detection and frequency modulation. In present work, the phase detection images were used. Generally, the magnetic tip is magnetized along its longitudinal axis by a permanent magnet prior to each measurement and the axis always keeps perpendicular to
the surface of film during scanning. The MFM detects the stray field of the film and consequently is sensitive to the magnetization perpendicular to the surface but insensitive to the in-plane component of magnetization \[10\]. The Bruker MESP probes coated by a Co/Cr bilayer were used, and the lift height was set to be 50 nm for all the measurements. To avoid the possible perturbation between the magnetic tips and the domains of specimens, each MFM picture was recorded twice to ensure reliability \[32\]. All the MFM characterizations were performed at room temperature in air.

3. Results

3.1. Crystal structures and chemical composition of Co thin films

Figure 1(a) shows an image of as-deposited Co thin film with a thickness of 100 nm on Si substrate captured by a digital camera, which shows that the centimeter scale film is uniform and well adhered to the substrate, without any macroscopic crack, void, or interfacial delamination. Figure 1(b) shows the XRD spectrums for the 100 nm Co films before and after annealing at 300 °C. The Si (333) peak appears at \(2\theta = 58.3^\circ\) (JCPDS 39-0073) for all the specimens. The as-deposited films only exhibits a weak peak at \(2\theta = 84.2^\circ\) (JCPDS 05-0727) corresponding to the Co (1013) texture, manifesting the as-deposited Co films are of weak crystallinity and may consist of very small grains \[8\]. After annealing, besides (1013) peaks, the XRD spectra of Co films show a sharp and fairly intense diffraction peak at 44.7° from crystalline plane (0002) (JCPDS 05-0727), indicating their highly crystalline nature. The dominant (0002) peaks reveal the annealed cobalt films are mainly composed of hexagonal close packed (HCP) phase with the c-axis preferentially oriented perpendicularly to the film surface \[10, 11\]. By comparing the XRD spectra before and after annealing, it can be inferred that the HCP phase is formed during annealing of Co films. As we will show below, the perpendicular magnetic anisotropy of annealed Co films is closely related to this HCP phase. It is worth noting that the analysis of the diffraction patterns in XRD is based on the standard JCPDS cards. In addition to JCPDS 05-0727, the available standard JCPDS cards for Co films also include: JCPDS 01-1278 and JCPDS 01-1277. It should be emphasized that even based on any of these JCPDS cards, we can draw the same conclusion that the HCP phase is developed after the 100 nm-thick Co films annealed at 300 °C.

The main panel in figure 1(c) shows the typical EDS spectrum of a 100 nm Co film annealed at 300 °C, and each peak is labeled with its corresponding element. The Co films with the same thickness of 100 nm regardless of annealing have a similar EDS pattern. Inset image in figure 1(c) is an EDS mapping at the edge of a Co film.
sample with green color representing the intensity of Co Kα1 signal. From the EDS spectrum, we find that in addition to cobalt, it also contains Si signals from the underlying substrate and C from SEM chamber (e.g., carbonaceous contaminants [33], amorphous carbon induced by electron-beam radiation [34]). And a little O signal may be from the native oxide of both silicon and cobalt. The Co films were well protected from further oxidation during the post annealing process under hydrogen atmosphere. The atomic percent of oxygen in Co films thus yield a slight decrease from 0.62% (for as-deposited) to 0.52% (annealed), as shown in the table (inset of figure 1(c)).

3.2. Morphology of Co film with various thicknesses before and after annealing at 300 °C
Figure 2 illustrates SEM images of the Co films with thicknesses of 8 nm and 25 nm on Si substrate before and after annealing. The as-deposited 8 nm Co film is continuous, however, some voids and cracks appeared in a few localized regions of the film surface (figure 2(a)). For cobalt film, even if the thickness is smaller, it can be continuous. Myers et al had reported that the cobalt thin film was still continuous when its thickness was reduced to 4 nm, but layers of 2 nm thickness didn’t form a fully continuous film [35]. The 25 nm-thick film exhibits a smooth surface without discernable defects (figure 2(b)). After annealing, the morphology of Co films was changed, as shown in figure 2(c) and (d). The annealed 8 nm film is full of microcracks and nanoparticles. The incremental density of microcracks may be attributed to the growth of crystalline grains, release of internal stress and/or shrinkage of Co thin films on the silicon surface [26–28]. Compared with the 8 nm film, the annealed cobalt film of 25 nm thickness remains continuous although there is some aggregation of cobalt nanoparticles on its surface. For thicker Co films, such as 50 nm and 100 nm, there is no significant change in the morphology before and after annealing from the SEM characterization (SEM images not shown).

Figure 3 shows the topographic images of the Co films with thicknesses varying from 8 nm to 100 nm before and after annealing. Notice that except for the first column, the two images in each column were captured from the same region. All the images in figure 3 are of the same size (2 μm × 2 μm). From the observation of atomic force microscope (AFM), it is shown that for the as-deposited 8 nm Co film, a smooth surface with few particles can be clearly observed (figure 3(a)), whereas the 100 nm-thick film is covered with dense nanoscale particles with an average height and diameter of 4.2 nm and 20.9 nm, respectively (as shown in figure 3(d), calculated by NanoScope analysis software). For thinner films (e.g., 8 nm), there are a lot of nanoscale pits or cracks on the surface after being annealed (figure 3(e)), whereas for thicker films (e.g., > 25 nm), the surfaces get smoother. The average height and diameter of the particles of 100 nm-thick annealed films are reduced to 2.3 nm and...
20.5 nm, respectively (figure 3(h)). There are several individual particles (marked by red arrows in figures 3(c) and (d)) whose shapes and sizes barely varied after annealing (figures 3(g) and (h)). The reason may be that these nanoparticles are stable even if after annealing at 300 °C. Perhaps higher annealing temperature and longer annealing time can make these particles smaller and the surface smoother [11, 14].

With the help of NanoScope analysis software, the root-mean-square (RMS) surface roughness can be obtained from AFM topographic images and thus a quantitative morphological analysis of Co films could be carried out [10]. Figure 4 shows the dependence of RMS roughness on the film thickness. To reduce the errors, the roughness value of each specimen is obtained from at least ten topographic images mapped in different areas. As shown in figure 4, the RMS roughness of as-deposited films increases almost linearly with incremental thickness of film. The roughness of the Co film surface is slightly reduced after annealing, especially for 100 nm films (the RMS variation is about 0.5 nm), which is consistent with the AFM observation presented above. The slight reduction of RMS roughness by annealing may be due to the relaxation of stresses, the growth of grains resulting in more compact in the Co film [26–28]. Note that the roughness of annealed 8 nm film is not shown in figure 4, because the RMS measurement cannot avoid the error caused by the pseudo-image induced by defects such as mass cracks. The RMS roughness of silicon is 0.32 nm ± 0.2 nm, as shown by a gray line. The roughness of as-deposited 8 nm film is close to that of the substrate, indicating that RMS of ultrathin films is mainly determined by the substrate.
3.3. Magnetic domains of Co film with various thicknesses before and after annealing at 300 °C

The magnetic domain, uniformly magnetized region, is a key element of magnetic microstructure [32]. Magnetic domain structures of the Co films before and after annealing were visualized by magnetic force microscopy (MFM) as shown in figure 5, where the two images in each column were captured from the same region (except 8 nm-thick samples). All the MFM pictures are of the same size (2 μm × 2 μm). The domain configurations of as-deposited Co films with different thicknesses are illustrated in the first row of figure 5. The MFM images show that the as-deposited films have irregular shaped domains with fuzzy boundaries, which may be derived from an in-plane component of magnetization, in-plane stray field or/and non-coplanar magnetization textures of the films [4, 11, 32].

The magnetic domain structures of the annealed Co films are displayed in the second row of figure 5. After being annealed at 300 °C, a stripe-like magnetic domain structure can be observed in the Co films with thicknesses of 50 nm and 100 nm (figures 5g and h)). The larger the thickness of cobalt film, the clearer the domain configuration. This observation shows that the stripe-like domains can be obtained by a thermal annealing process. However, even after annealing, the MFM images of Co films with thickness less than 25 nm do not change much and the stripe-like magnetic domain structure still does not appear (figures 5e and f)), which elucidates the effect of annealing on the magnetic domain depends on the thickness of the cobalt film. In these stripe domain regions, the out-of-plane magnetization plays a more decisive role than the in-plane magnetization. The white and black contrast represents striped domains with alternating up and down magnetization [7]. The minimum thickness (hereafter called the critical thickness) of annealed cobalt films having striped domain configurations may range between 25 nm and 50 nm, well coinciding with the experimental values in Brandenburg et al [21]. If the film thickness is larger than the critical one, the magnetization direction in the film may rotate from in-plane to out-of-plane. By comparing the magnetic domain of 50 nm and 100 nm Co films before and after annealing, it is found that the perpendicular magnetic anisotropy is markedly enhanced by annealing. Moreover, the discernable domain boundaries in figure 5(h) are the Bloch type walls wherein the magnetization rotates parallel to the wall surface with a demagnetizing field on the film surface to reduce the total magnetostatic energy of the film [14]. As shown in the MFM images (figure 5 (h)), the annealed 50 nm and 100 nm Co films exhibit a periodic stripe domain patterns with width, d, which can be used to calculate the energy density of domain wall (will be described in the discussion section) [10–13].

3.4. Magnetic domains for 50 nm-thick Co films annealed at different temperatures

We also examined the influence of annealing temperature on the magnetic domain of Co films. As we mentioned above, the as-deposited 50 nm-thick Co films have no clear domain structure. Therefore, we annealed the as-deposited Co film samples with the same thickness of 50 nm on SiO₂ substrate at 200 °C, 400 °C and 500 °C, respectively. Figure 6 shows MFM images of both as-deposited and annealed Co films. It is found that periodic stripe-like domain structures can be observed in the all films annealed at temperatures ranging from 200 °C to 500 °C. Much attention should be paid to the case of the film annealed at the temperatures > 400 °C. It had been reported that a phase transition from hexagonal closed packed (HCP) to face centered cubic (FCC) occurred at 350 °C [27, 28], which would weaken the perpendicular anisotropy of the
Co films [3]. However, the magnetic domains of the Co films annealed at higher temperatures (e.g., 400 °C and 500 °C) exhibit stripe-like pattern, and showed no significant difference from that annealed at 200 °C. The reason is that the phase transition between HCP and FCC is reversible [36]. During annealing, the transformation from HCP to FCC occurs in the heating process, but the reverse transformation from FCC to HCP takes place in the subsequent cooling process. It can be confirmed from the XRD spectra that the Co films are all HCP-phase at room temperature. Even if FCC is present during the annealing process (at temperatures >350 °C), it has little effect on the magnetic domain of the Co films at room–temperature.

4. Discussion

We first aim to figure out the correlation between the film morphology and magnetic domain structure. As shown in figures 3 and 5, the morphology and magnetic domains of cobalt films with various thicknesses are mapped in the same region (except for 8 nm film). The surface of cobalt film is not smooth, and there are a lot of nanoparticles (for 25 nm ~ 100 nm films) or some pits or microcracks (for 8 nm film). In the MFM images, however, we did not observe any characteristic of the nanoparticles in the corresponding AFM topographic images, even the larger nanoparticles (as marked by the arrow in figures 3(c) and (d)), indicating that these nanoparticles do not affect the magnetic stray field distribution on the surface of the Co film. Moreover, although there are many nanoscale pits or cracks on the surface of the annealed 8 nm-thick cobalt film, no pit or crack trace is observed in the corresponding MFM images, which may be due to the fact that these pits or cracks are so narrow that they are still covered by the profile of adjacent stray fields [30]. It has been reported that the surface morphology of cobalt films can affect their own domain structures [37], however our observation shows that this effect is not significant. This difference may be due to the different thickness of the films used. In [32, 37], the thickness of Co/Au (111) was only 2 ~ 4 monolayers (ML), while in our case the films in the thickness range of 8 nm ~ 100 nm.

We then discuss the relationship between the thickness and its domain structure of the Co thin film. The feature of the domain structure of magnetic materials depends on the relative magnetic anisotropy constant Q, which is defined as the ratio of the uniaxial anisotropy constant \(K_u\) to the stray field energy constant \(K_d = M_s^2 / 2\mu_0\), where \(M_s\) is the saturation magnetization and \(\mu_0\) is vacuum permeability [11, 12]. For HCP phase cobalt films, the Q value is 0.4 [8, 11, 12], which means the uniaxial magnetocrystalline anisotropy is not strong enough to exclusively dominate the magnetization orientation but other anisotropies, such as shape anisotropy and surface anisotropy, are also determinative. As a result, it is generally believed that both the magnetization direction and domain structure depend on the thickness of the film [3–5, 7–11, 21]. If the thickness of the film is reduced below a critical thickness, the magnetization aligns spontaneously parallel to the film surface due to the shape-induced anisotropy, whereas the thickness exceeds a critical value, the magnetization gradually turns to the direction perpendicular to the film surface [7, 37–39]. What is the critical minimum thickness of cobalt film? Recently, some theoretical and experimental studies have been carried out...
on this problem. For example, the critical thickness for (0002) HCP cobalt films grown by molecular beam epitaxy (MBE) sandwiched between two Ru thin films occurs between 10 nm and 50 nm [7], and the (0002) HCP Co nanocrystalline thin films deposited by thermal evaporation is between 50 nm and 60 nm [9, 10]. A summary of the reported critical thicknesses of cobalt films is shown in Table 1 [3–5, 7–10, 21]. From the data in the table, we can see that although they are all cobalt films, the critical thickness apparently varies with the deposition method, the processing conditions, the substrate material and its crystal orientation. Most of the measured critical thicknesses are larger than that of the theoretical prediction (i.e., 20 nm or 30 nm) [6, 21].

| Critical Thickness | Substrate | Preparation Method | Crystallography |
|--------------------|-----------|--------------------|-----------------|
| This work          | 25 ~ 50 nm (annealed) | (111) Si | EBE | Strong (0002) HCP |
|                    | >100 nm (as-deposited) | Ru/(012) sapphire | MBE | Poor crystallinity & weak (013) peak |
| Hehn et al[7]      | 10 ~ 50 nm | (100) Si & glass | Evaporation | Dominant (0001) HCP |
| Kharmouch et al[8] | 125 ~ 173 nm | (100) Si | Evaporation | Good (0001) HCP (thick film) |
| Chérif et al[3]    | 70 ~ 125 nm | Glass | MOCDV | Strong (0002)HCP/(111) FCC |
| Donzelli et al[5]  | 43 ~ 86 nm | (111) Si | MS | (0001) HCP |
| Brandenburg et al[21] | 40 ~ 45 nm | Ru/(0001)Al2O3 | PLD | Pronounced (0001) HCP |
| Kozowski et al[10, 11] | 36 nm | Glass & NaCl | TE | Mainly composed of (0001) HCP |

1 A critical thickness estimated using a model in Saito et al[40] by Hehn et al[7] was 20 nm.
2 The critical thickness was speculated to be about 50 nm.
3 Chérif et al[3] had proved weak crystallinity for thin films.

Table 1. Critical thickness experimentally obtained in this work and seven different previous work with the corresponding substrate, preparation method and crystallography of Co thin films.
we need to calculate the domain width for the method with over 1000 intercepts counted for each MFM images 70.3 nm different positions on the surface of annealed 100 nm-thick Co evaporation and annealed at atmospheric pressure in a gas mixture 5. Conclusions

This formula shows that the dependence of domain width \( d \) on the film thickness \( (T) \) is as follows: \( d \propto \sqrt{T} \). In order to obtain \( \sigma_w \), we need to calculate the domain width first. The domain width can be directly obtained from the domain pattern in MFM images by using standard linear intercept method \([11, 13, 14, 32, 45]\). We used this method with over 1000 intercepts counted for each MFM images \((2 \mu m \times 2 \mu m)\), which have been mapped at different positions on the surface of annealed 100 nm-thick Co film. The domain width is measured to be 70.3 nm \( \pm 5.2 \) nm. For \( M_s = 1.8 \) T \([21]\), \( T = 100 \) nm. Therefore, the energy density of domain wall estimated by formula (1) is 17.3 mJ m\(^{-2}\) \( \pm 3 \) mJ m\(^{-2}\). This value is very close to the theoretical prediction of 14.2 mJ m\(^{-2}\) (for an isolated Bloch wall, \( \sigma_w = 4\sqrt{A\lambda}, \) where \( A \), exchange constant, is \( 2.8 \times 10^{-11} \) J m\(^{-3}\) and \( K_u = 4.5 \times 10^5 \) J m\(^{-3}\) \([21, 32]\)). Furthermore, using the same estimation method, we calculate \( \sigma_w \) for the 50 nm-thick Co films annealed at different temperatures to further explore the effect of annealing temperature on the energy density of domain wall. The estimation results including domain wall energy density \( (\sigma_w)\) and domain width \( (d) \) are shown in figure 7. It’s found that the domain-wall energy density for 50-nm-thick films is larger than that for the 100 nm-thick ones, which may be due to the stronger in-plane component in the thinner cobalt films. However, the domain wall energy density of the cobalt film annealed at different temperatures does not change much (figure 7), manifesting the annealing temperature has no significant effect on the energy density of domain wall. This result is consistent with that the domain wall energy density depending only on the exchange constant and uniaxial anisotropy constant of the film, and the exchange constant is generally temperature dependent and the uniaxial anisotropy constant is mainly related to the crystal structure as well as the thickness for thin film samples \([32]\).

5. Conclusions

Centimeter scale cobalt films with thicknesses from 8 nm to 100 nm have been prepared by electron beam evaporation and annealed at atmospheric pressure in a gas mixture flow of hydrogen and argon. For the as-deposited Co films, they are uniform and continuous (except for 8 nm film) and their surfaces are covered with nanoscale particles. The observation of SEM and AFM images suggest that there is no obvious change in the morphology before and after annealing especially for thicker films (such as 50 nm and 100 nm). The root-mean-square (RMS) roughness of the Co film surface, however, is slightly reduced after annealing. The annealed films are mainly composed of hexagonal close packed (HCP) phase of cobalt with the c-axis preferentially oriented normal to the film surface.

The magnetic domains of the Co films were visualized by magnetic force microscopy (MFM). It is found that the as-deposited films have irregular domains with fuzzy boundaries. After annealing, a periodic stripe-like magnetic domain can be formed in the Co films with a thickness more than 50 nm and a perpendicular magnetic anisotropy (PMA) of film is obviously enhanced. For 50 nm-thick films, a stripe domain can be obtained by annealing at temperature as low as 200 °C. However, even after annealing, the domain structures of Co films with thickness < 25 nm remain unchanged and the stripe-like domain pattern still does not appear, which indicates that the effect of annealing on the magnetic domain also depends on the thickness of the Co films. The critical thickness of Co films for the onset of PMA is greatly reduced from \( \sim 100 \) nm (as-deposited) down to 25 nm \( \sim 50 \) nm (annealed). The main reason for the formation of stripe domains (also enhancement of PMA) in Co films is that the annealed Co films are composed of HCP phase with a strong texture along (0002) direction. From the periodic stripe magnetic domains, the domain-wall energy density of the annealed Co film

\[
\sigma_w = \frac{\sigma_w T}{0.136M_s^2}
\]
on the silicon surface was theoretically estimated to be about 17.3 mJ m$^{-2}$ ± 3 mJ m$^{-2}$ and the annealing has almost no effect on the energy density of the domain wall.

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