Effect of carbon additive on microstructure evolution and magnetic properties of epitaxial FePt (001) thin films

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\textbf{A B S T R A C T}

FePt:C thin films were deposited on CrRu underlayers by DC magnetron co-sputtering. The effects of C content, FePt:C film thickness and substrate temperature on the microstructural and magnetic properties of the epitaxial FePt (001) films were studied. Experimental results showed that even with 30 vol.\% C doping, the FePt films could keep a (001) preferred orientation at 350 °C. When a FePt:C film was very thin (<5 nm), the film had a continuous microstructure instead of a granular structure with C diffused onto the film surface. With further increased film thickness, the film started to nucleate and formed a columnar microstructure over continuous FePt films. A strong exchange coupling in the FePt:C films was believed to be due to the presence of a thin continuous FePt layer attributed to the carbon diffusion during the initial stage of the FePt:C film growth. Despite the presence of a strong exchange coupling in the FePt:C (20 vol.\% C) film, the SNR ratio of the FePt:C media was about 10 dB better than that of the pure FePt media. The epitaxial growth of the FePt:C films on the Pt layers was observed from high resolution TEM cross sectional images even for the films grown at about 200 °C. The TEM images did not show an obvious change in the morphology of the FePt:C films deposited at different temperatures (from 200 °C to 350 °C), though the ordering degree and coercivity of the films increased with increased substrate temperature.

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1. Introduction

(001) textured L1\textsubscript{0} FePt films with a large magnetocrystalline anisotropy above 10\textsuperscript{7} erg/cc have been considered as the most promising candidate for ultrahigh density (beyond 1 Tbit/in\textsuperscript{2}) magnetic recording media [1]. Usually as-deposited FePt alloy thin films are a disordered fcc (face-centered-cubic) phase and tend to show a (111) texture. High temperature (above 550 °C) post annealing is needed to form the L1\textsubscript{0} phase [2,3]. In order to use FePt films as high density perpendicular magnetic recording media, it is necessary to fabricate chemically ordered L1\textsubscript{0} FePt (001) films with uniform and well dispersed nano grains at relatively low temperatures on glass or AI substrates. In the past few years, significant progress has been made to lower the ordering temperature and control the orientation and grain size of FePt films. Various underlayers or substrates, such as MgO [2], Ag [4], CrRu [5] and CrMo [6] have been successfully used to promote the L1\textsubscript{0} FePt (001) texture for perpendicular magnetic recording. The decreases of grain size and exchange coupling are usually achieved by adding a third material into the films. Materials such as Al\textsubscript{2}O\textsubscript{3} [7], Al\textsubscript{2}O\textsubscript{3} [8], AlN [9], TaN [10], Ag [11], MgO [12], W and Ti [13] have shown to be promising in controlling the FePt grain size in ordered films. Other studies involving FePt with Ni, B, BN, or C [14,15] as additives have demonstrated that these can reduce the magnetic coupling. However, for practical application of FePt as a perpendicular magnetic recording media, one basic requirement for choosing a third element or alloy as an additive is to keep the (001) preferred orientation in the FePt films after doping. In our previous work, the easy axis of FePt films grown on CrRu and CrMo underlayers changed from perpendicular to longitudinal by Ag [16] and Cu [17] doping. Though the reasons why the orientation changed after Ag and Cu doping are still not very clear yet, the suppressed mobility of Fe and Pt adatoms should be one important reason.

The atomic radius of Carbon (~0.077 nm) is much smaller than those of Pt (0.138 nm) and Fe (0.126 nm). Therefore, it is very likely that carbon atoms can diffuse through interstitial sites as in the case of diffusion in CoCr based alloys [18] and the FePt films can epitaxially grow on the underlayers. Though the atomic radius of Boron (0.097 nm) is also much smaller than those of Pt and Fe, FePt:B longitudinal media showed improved SNR and nonlinear...
transition shift (NLTS) compared to FePt media due to a reduced inter-granular exchange coupling [14]. Boron can form alloys with Fe, which is not favorable for grain size reduction. In addition, a B additive in a sputter target may cause some additional defects in the films during sputtering, since it is not electrically conductive. Ko et al. reported that the L10 ordering and grain growth kinetics of FePt (001) films could be finely controlled by C doping [19]. However, the use of MgO single crystal substrates limited the practical applications of FePtC films. Yan and co-workers reported a synthesis of L10 FePt-C (001) composite films by multilayer deposition followed by a subsequent thermal annealing [20], but a high temperature (550 °C) long time post annealing was not compatible with the current manufacturing processes for hard disks. In our previous work, flyable L10 FePt-C composite perpendicular magnetic recording media sputtered on glass substrates at 350 °C have been demonstrated [21]. However, a thin continuous FePt film under FePt-C granular film has been found, which results in a strong exchange coupling. To develop well isolated FePt-C nanocomposite granular media, it is necessary to understand the growth mechanism of C doped FePt films. In this paper, the effect of C additive, FePt-C film thickness and substrate temperature on the microstructure evolution and magnetic properties of the epitaxial FePt films is addressed.

2. Experiments

A multilayer structure of Glass/Cr90Ru10 30 nm/Pt 2nm/FePt:C were prepared by DSI home designed sputtering system with a base pressure better than 1 × 10⁻⁸ Torr. The FePt:C films were prepared by co-sputtering FePt alloy target and C target. The thicknesses of Cr90Ru10 and Pt layers were fixed at 30 nm and 2 nm, respectively, while the thickness of the FePt:C films was varied from 5 to 30 nm. The C content (x) was varied from 0 to 30 vol.%. The working pressures were 3 mTorr, 10 mTorr and 10 mTorr for the Cr90Ru10, Pt, and FePt:C layers, respectively. The deposition temperature for the CrRu and Pt layers was fixed at 350 °C and that for the FePt:C layers was varied from 200 to 350 °C. The crystallographic texture of the films was examined with X-ray diffraction (XRD) using a Cu Kα radiation. The microstructure of the films was characterized with transmission electron microscopy (TEM). The magnetic properties of the FePt-C films were measured using a Polar Kerr and a vibrating sample magnetometer (VSM). A Guzik spin-stand (1701B) with a commercial ring head (for longitudinal media) was used to roughly characterize the recording properties of the fabricated FePt:C media.

3. Results and discussion

3.1. Influence of C content

In this section, the thicknesses of the Cr90Ru10 underlayers, Pt buffer layers and FePt:C films are 30 nm, 2 nm and 10 nm, respectively. The deposition temperature for all the layers is fixed at 350 °C. The C content is varied from 0 to about 30 vol.%. Fig. 1 shows the XRD spectra of the FePt:C films with various C volume fractions deposited at 350 °C. From the XRD spectra, one can see that all the peaks are attributed to the Cr90Ru10 underlayers and FePt films, while no peaks are from carbide phases, which indicates that most C remains as a pure element matrix. With increasing C fraction from 0 to 30 vol.%, the FePt (001) peak locations remain almost unchanged, illustrating that the lattice constant c has no change. The FePt (002)+(200) peaks of the samples doped with 5, 10 and 15 vol.% C locate at a slightly higher angle than that of the pure FePt film, indicating an increase of the FePt (002) content. With further increase of the C content up to 30 vol.%, the FePt (002)+(200) peak shifts to a lower angle, indicating an increase of the fcc FePt (200) content. The XRD results depict that the FePt films can keep a (001) preferred orientation after incorporating C up to 20 vol.%, which is different from those doped with Ag [16] or Cu [17]. In some
previous reports, the easy axis of FePt films grown on CrRu or CrMo underlayers changed from perpendicular to longitudinal by Ag [24] and Cu [25] doping, which the suppressed mobility of Fe and Pt adatoms by Cu and Ag doping was believed to be a main reason. The atomic radii of Cu and Ag atoms are comparable with those of Fe and Pt atoms, and thus some Fe and Pt atoms have to overcome the energy barriers caused by the Cu and Ag atoms before reaching their minimum energy sites. As for the C additive, the atomic radius of carbon atom (0.077 nm) is much smaller than those of Pt (0.138 nm) and Fe (0.126 nm) atoms. Thus, it is likely that the C atoms can diffuse through interstitial sites as in the case of CoCr based alloys [18] and the mobilities of the Fe and Pt adatoms are not restricted so much. Thus, most of the Fe and Pt adatoms can move to their minimum energy sites and align with the atomic arrangement of the underlayer.

Fig. 2 shows the texture quality of the FePt:C films characterized by the FePt (001) peak rocking curves. The FWHM of the curves increases slightly from about 7.2° to 9.2° with increased C content from 0 to 30 vol.%, which implies that a relatively good (001) texture can be maintained even after incorporating C up to 30 vol.%. The chemical ordering degree, \((I_{001}/I_{002})^{1/2}\) \((I_{001} and I_{002} are the intensities of FePt (001) and (002) peaks, respectively), as a function of C content is shown in Fig. 3. \((I_{001}/I_{002})^{1/2}\) first increases with increasing C content from 0 to 15 vol.%, and then decreases with further increasing C volume fraction. This signifies that a small amount of C doping can slightly enhance the ordering degree, while a large amount of C doping can decrease the ordering degree. These results are similar to those reported previously [19]. The variation of the ordering degree of the FePt films with increased C content may be explained as follows. When the C content is small, C atoms may diffuse through FePt interstitial sites during growth. The diffusion of C atoms may provide a driving force for a FePt phase transformation from fcc to fct. As the C content is large, the epitaxial growth of the FePt on the Pt buffer layer may be suppressed, and the FePt/Pt interface may be destroyed by the C adatoms. The lattice mismatch induced epitaxial stress may release, thus the FePt films have a lower ordering degree. The diffusion of C during deposition has been verified by TEM images and will be discussed in the following section.

The microstructural properties of the FePt:C films are investigated by TEM. Figs. 4 (a) and (b) show the plan-view TEM images of the FePt:C films with 10 and 20 vol.% C, respectively. Under this sputter condition, pure FePt films grown on Cr90Ru10 underlayers show a continuous microstructure [22]. When 10 vol.% C is doped, a granular structured film is formed though some grains aggregate together and not all grain boundaries can be distinguished, and the grain size roughly follows a Gaussian distribution with a mean diameter of about 9.9 nm and a standard deviation of 2.3 nm. As the C content increases to 20 vol.%, the TEM image shows well-defined grain boundaries and relatively uniform grain sizes. The mean grain size and standard deviation are about 5.6 nm and 1.6 nm, respectively. In the both cases, the grain size distribution of the FePt:C films is relatively narrow and encouraging with reference to the results reported previously [19].

Figs. 6–7 show both the low magnification and the high-resolution cross-sectional images of the FePt:C media doped with 10, 20 and 30 vol.% C, respectively. Fig. 5 shows the TEM images of a 15 nm thick FePt:C films. All the images in Figs. 5–7 reveal that the films have three layers, according to different contrast. The top FePt:
C layers show a granular microstructure with clear grain boundaries, confirming that the FePt:C films have well-defined grain boundaries and uniform grains, while the Cr$_{90}$Ru$_{10}$ and Pt layers show a continuous microstructure. In all the images, the middle continuous layers are about 7 nm thick which is much thicker than 2 nm from the design, while the top FePt:C layer is thinner than the designed thickness. It has been reported that the Cr diffusion could be effectively blocked by a thin Pt buffer layer [23,24]. Thus, the middle continuous layers should consist of a 2 nm thick Pt layer and a 5 nm thick FePt continuous layer. This may be explained as follows. It is well known that an epitaxial stress is very strong when a film is very thin. Thus, at the initial growth stage of the FePt:C film, a strong epitaxial stress governs the thin film growth mechanism: i.e. the Fe and Pt adatoms can follow the atomic arrangement of the Pt layer and then a continuous microstructure is formed. The C adatoms may diffuse onto the surface of the FePt layer at the same time. With the increase of the FePt:C film thickness, the epitaxial stress releases. More C atoms on the FePt film surface may inhibit the mobility of the Fe and Pt adatoms and reduce the surface bonding of the FePt film. Thus, some Fe and Pt adatoms cannot reach their epitaxial sites, but instead forming three-dimensional islands. The ball-like shape of the FePt grains should be mainly governed by the film surface tension. That the FePt grains can have a (001) preferred orientation is mainly due to the epitaxial growth as observed from the high-resolution images [Figs. 5 (b), 6 (b) and 7 (b)]. An interesting phenomenon is that the morphology of the FePt layers changes from column-like to ball-like with increased C content, which should be due to the surface tension and deteriorated epitaxial stress resulted from the lattice misfit at the Pt/FePt interface.

There exists a light gray contrast layer (~2 nm) between the middle continuous and top granular layers as seen in the low magnification images [Figs. 5 (a), 6 (a) and 7 (a)]. This layer has been generally considered a “initial layer” caused by inter-diffusion between two layers [25] and/or by elastic strain due to a lattice misfit between two layers [26]. In this study, the continuous lattice images from the Pt to the FePt continuous layers and then to the FePt:C granular layers are observed [Figs. 5 (b), 6 (b), and 7 (b)], which indicates the epitaxial growth between Pt and FePt/FePt:C layers. In addition, some amorphous areas are also observed between the Pt/FePt and FePt:C layers [Figs. 5 (b) and 6 (b)], which should be carbon atoms. Thus, the regions with a light contrast between the continuous FePt layer and overlying granular FePt grains should be mainly due to the carbon. For the FePt:C film doped with 30 vol.% C, the high resolution image does not show a clear continuous lattice from the Pt/FePt layer to the FePt:C layer, but a rather poor texture which is believed to be caused by the C additive [marked with a circle in Fig. 7 (b)].

Fig. 5. TEM cross-sectional images of the 15 nm thick FePt:C film doped with 10 vol.% C: (a) low magnification and (b) high resolution.

Fig. 6. TEM cross-sectional images of the 10 nm thick FePt:C film doped with 20 vol.% C: (a) low magnification and (b) high resolution.
only in the film normal direction. The evolution of the microstructure with increased film thickness should obey the minimum system energy principle. In our case, the thickness dependence of the microstructure should be mainly attributed to the competition between the epitaxial stress, surface tension and surface energy.

Fig. 10 shows the out-of-plane hysteresis loops of the FePt:C films doped with various C contents as measured by polar Kerr. The loops illustrate that all the FePt:C films have a perpendicular magnetic anisotropy, confirming that the easy axis is perpendicular to the film plane, which is consistent with the preceding XRD analysis. Table 1 summarizes the out-of-plane coercivity \( H_{\text{c,perp}} \), nucleation field \( H_{\text{N,perp}} \) (field where \( M \) starts to drop from saturation \( (M=0.95\text{Ms}) \)), magnetization squareness \( S \) \( (5\%=(1-\text{Mr}/\text{Ms})/\text{Slope at } H_{\text{c,perp}}) \) of the FePt:C films. From Table 1, one can see that \( H_{\text{c,perp}} \) increases from about 2.5 kOe to 3.9 kOe with increasing C content from 0 to 15 vol.%, and then decreases with further increased C content, showing a similar trend to that of the anisotropy field. When the C content is ≤20 vol.%, relatively large negative \( H_{\text{N,perp}} \) and high \( S \) are obtained, both of which (large negative \( H_{\text{N,perp}} \) and high \( S \)) are required for lowering the DC noise in perpendicular media. The \( S \), which is usually used to reflect the exchange coupling in perpendicular magnetic recording media, does not change much with increased C content, indicating that the exchange coupling does not change much with increased C content either. For all the samples, \( S \) is higher than 0.85, indicating a strong exchange coupling in all the FePt:C films. These results are similar to those reported by Yan and co-workers [20]. It is well understood that inter-granular exchange coupling depends on the microstructure of magnetic films. Normally, a continuous magnetic film always shows a strong exchange coupling and an isolated granular magnetic film shows a weak exchange coupling. In our case, the strong exchange coupling in the FePt:C films should be due to a continuous FePt film between Pt layer and FePt:C granular layer. Because the FePt grains in the continuous layer have a strong exchange coupling, this continuous layer also strongly couples with the overlying granular FePt grains. Thus, the magnetization of the granular FePt grains switches together with the continuous FePt layer when an external magnetic field is applied.

In order to confirm the proposed growth mechanism of the FePt:C films, the angular dependence of coercivity is measured by VSM. Fig. 11 shows the angular dependence of the coercivity profiles of the FePt:C films with 0, 10, 20 and 30 vol.% C doped. The magnetization reversal of the pure FePt films grown on the CrO/CoRu/C underlayers follows a typical domain-wall motion model. The angular dependence of the \( H_{\text{c,perp}} \) profiles of the FePt:C films shows that the magnetic reversal mechanism deviates from the domain wall motion predominant model and tends to approach to a rotation predominant model with increasing C content. The variation of the

Fig. 7. TEM cross-sectional images of about 10 nm thick FePt:C film doped with 30 vol.% C: (a) low magnification and (b) high magnification.

Fig. 8. Proposed two-step growth model for FePt:C films. (a) Step 1: FePt is epitaxially grown on CrO/CoRu/pt and carbon is squeezed onto the top surface of FePt film. (b) Step 2: With increased film thickness, carbon is accumulated to some extent, and the epitaxial stress is released. Fe and Pt atoms nucleate on some sites and then the nuclei grow larger.
magnetic reversal mechanism should be due to the variations of the microstructures of the FePt:C films. As discussed above, the pure FePt films have a continuous microstructure, while the FePt films doped with carbon consist of an FePt:C granular layer and a thin continuous FePt layer. Thus, the pure FePt films show a domain wall motion dominant magnetic reversal mechanism while the FePt:C films show a combined domain wall motion and rotation mode.

A Guzik spin-stand (1701B) with a commercial ring head (for longitudinal media) was used to roughly characterize the recording properties of the fabricated FePt:C media. For the media used for read-write testing, the FePt:C films were deposited on commercial HOYA 2.5 in. glass substrates. The deposition temperature for the Cr90Ru10, Pt, and FePt:C layers was fixed at 350 °C. Finally, a 5 nm thick carbon overcoat was deposited on the top of the FePt:C films at room temperature when the samples were cooled down to room temperature in vacuum. Only selected interesting media were measured.

The reproduced waveforms of the individual tracks of the FePt:C perpendicular media doped with 20 vol.% C at various densities are shown in Fig. 12. The square waveforms are observed at a linear density of 59 kfci, indicating that it is a perpendicular medium, though the peaks are not smooth and there exist some small fluctuations. It is well understood that the demagnetization field is quite large at a low linear density of 59 kfci, indicating that it is a perpendicular medium, though the peaks are not smooth and there exist some small fluctuations. It is well understood that the demagnetization field is quite large at a low linear density of 59 kfci, indicating that it is a perpendicular medium, though the peaks are not smooth and there exist some small fluctuations. Thus, the fluctuations at the peaks should be mainly due to the demagnetization field and the inter-granular exchange coupling. As the linear density increases to 118 kfci, the read-back signals can still be distinguished. However, as the linear density is further increased to 221 kfci, some signals overlap in addition to the amplitude fluctuations, which should also be mainly due to the demagnetization field and strong inter-granular exchange coupling.

The SNRs with respect to the linear density for the pure FePt media and FePt:C media doped with 20 vol.% C are shown in Fig. 13. As seen from the graph, the SNR ratios of the FePt:C media doped with 20 vol.% C are about 10 dB better than those of the pure FePt media. It is worth noting that the SNR measurement is not absolute, but only relative. This is because the signal and noise are determined by the type of the head (e.g. sensitivity, writing capability) used, head-media spacing etc. As far as the recording media are concerned, for a given track width, smaller de-coupled grains will produce a higher SNR. In this study, the decreased grain size in the FePt:C media should be one of the main reasons for the improved SNR.

3.2. Influence of substrate temperature

In order to study the influence of $T_s$ on the microstructure and magnetic properties of the FePt:C films, a series of samples with a structure of Glass/Cr90Ru10 (30 nm)/Pt (2 nm)/FePt:C (15 vol.% C) (~10 nm) at different temperatures was fabricated. $T_s$ for the growing Cr90Ru10 underlayers was fixed at 350 °C, and for the Pt and FePt:C layers varied from 200 to 350 °C. The influence of the substrate temperature on the structural and magnetic properties of the FePt:C films are addressed below.

The crystalline orientation of the FePt:C media deposited at different substrate temperatures is shown in Fig. 14. At 200 °C, a
FePt (200) peak is dominant. The soft magnetic properties indicate that it is an fcc FePt (200) peak [Fig. 15]. In addition, a weak FePt (111) peak near 40° appears, indicating a poor texture in the FePt:C film. As the substrate temperature increases to 250 °C, a very weak FePt (001) peak appears, indicating a weak chemical ordering degree in the FePt:C film. In addition, the peak between 45° and 50° shifts to a higher angle, due to the appearance of a small amount of fct FePt (002) phase. The hysteresis loops and coercivity values illustrate that it is still an fcc FePt (200) predominant peak [Fig. 14]. With a further increase in substrate temperature, the intensity of the FePt (001) peak becomes stronger. In addition, the peak between 45° and 50° shifts to a higher angle again, depicting an improved fct FePt (002) phase. The TEM cross-sectional images of the FePt:C films reveal that the grain size does not rise with increased substrate temperature (from 200 to 350 °C) [Fig. 17], suggesting that the disorder-order transformation in these FePt:C films is a continuous type.

The magnetic properties of the FePt:C (15 vol.% C) films deposited at different substrate temperatures were measured by VSM. Fig. 15 shows both the in-plane and the out-of-plane hysteresis loops of the FePt:C (15 vol.% C) films grown on the Cr90Ru10 underlayers with different substrate temperatures. At 200 °C and 250 °C, the FePt:C films show the soft magnetic properties with a coercivity smaller than 300 Oe. At 300 °C and 350 °C, the films show a perpendicular anisotropy. The out-of-plane coercivity increases monotonously with

![Fig. 10. Out-of-plane hysteresis loops of FePt:C films with different C contents.](image-url)
increased substrate temperature [Fig. 16], which should be mainly due to the increased ordering degree. For the FePt:C (15 vol.% C) film deposited at 350 °C, there exists a "shoulder" in the second quadrant of the hysteresis loop [Fig. 15], indicating that there exists some magnetic soft phase in the film. A point worth noting is that the out-of-plane coercivity of the FePt:C (15 vol.% C) film measured by VSM (3.5 kOe) is smaller than that measured by polar kerr (3.9 kOe), which should be due to a slower field sweep rate during the VSM measurement, in addition to a different system sensitivity.

The microstructural properties of the FePt:C (15 vol.% C) films deposited at different substrate temperatures are investigated by TEM as shown in Fig. 17. The TEM cross-sectional images of the FePt:C films reveal that the average grain size of all the FePt:C (15 vol.% C) films is about 6 nm with increased substrate temperature from 200 °C to 350 °C.

<table>
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<th>C content (vol.%)</th>
<th>$H_{c,\perp}$ (Oe)</th>
<th>$H_n$ (Oe)</th>
<th>$S$</th>
<th>$S^\circ$</th>
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<td>0</td>
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<td>-2350</td>
<td>1</td>
<td>0.981</td>
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<td>10</td>
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<tr>
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<td>20</td>
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<td>1832</td>
<td>3697</td>
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<td>0.920</td>
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</tbody>
</table>

Fig. 11. Angular dependence of the coercivity profiles of the FePt:C films doped with 10, 20 and 30 vol.% C. (0° refers to the film normal direction).

Fig. 12. Reproduced waveforms of individual tracks at various densities for the FePt:C media doped with 20 vol.% C.

Fig. 13. SNR with respect to linear density for pure FePt media and FePt:C media doped with 20 vol.%.

Fig. 14. XRD spectra of FePt:C (15 vol.% C) films grown on Cr$_{69}$Ru$_{31}$ underlayers at different substrate temperatures.
350 °C. A continuous FePt layer is formed even for the sample deposited at 200 °C. To develop well exchange decoupled FePt film based perpendicular magnetic recording media, it is essential to eliminate the thin continuous FePt layer. From the epitaxial growth point of view, development of an underlayer with a well-isolated granular microstructure should be a key approach for the development of granular FePt media instead of only doping other elements into FePt films. The high resolution images show a clear continuous lattice from the Cr₉₀Ru₁₀ underlayers to the Pt buffer layers and further to the FePt:C layers, depicting that the Pt buffer layers first epitaxially grew on the Cr₉₀Ru₁₀ underlayers and then the FePt:C layers epitaxially grew on the Pt buffer layers during film deposition [Fig. 17(b), (f)]. It is worth highlighting that the epitaxial growth has indeed occurred at 200 °C [Fig. 17].

4. Conclusion

The influences of C content, FePt:C film thickness and substrate temperature on the microstructure evolution and magnetic properties of epitaxial FePt(001) films were studied. The following are the main conclusions drawn:

(1) The (001) oriented FePt:C films with perpendicular magnetic anisotropy and relatively narrow orientation distribution (rocking curve of FePt (001) <10°) were obtained with C doping.

(2) A small amount of C doping could slightly enhance the ordering degree, while a large amount of C doping could prohibit the ordering degree.

(3) A strong exchange coupling in the FePt:C media fabricated was due to the existence of a thin continuous FePt layer caused by the carbon diffusion during the initial stage of the FePt:C film growth. The continuous FePt layer should be mainly due to the continuous CrRu underlayer and the epitaxial stress suppressed the C diffusion. A two-step growth model for the FePt:C films was proposed.

(4) The SNR ratio of the FePt:C media doped with 20 vol.% C was about 10 dB better than that of the pure FePt media, despite the presence of a strong exchange coupling in the FePt:C film.

(5) The epitaxial growth of the FePt:C films on the Pt layers was observed from HR TEM cross sectional images even for the films grown at about 200 °C. The TEM images did not show an obvious change in the morphology of the FePt:C films deposited at different substrate temperatures (from 200 °C to 350 °C).

Fig. 15. Hysteresis loops of FePt:C (15 vol.% C) films deposited at different substrate temperatures.

Fig. 16. Coercivity with respect to substrate temperature for FePt:C films with 15 vol.% C.
though the ordering degree and coercivity of the films increased with increased substrate temperature.

References