Introduction

In recent years, in order to achieve a high recording density of over 1Tb/in² and conquer the super-paramagnetic size limit, the L1₀-ordered FePt film with high magnetization ($M_s \sim 1000$ emu/cc), large magnetic anisotropy energy ($K_u \sim 7 \times 10^7$ ergs/cc) and excellent corrosion resistance is considered as the most promising candidate for the media material in the heat assisted magnetic recording (HAMR) [1–3].

During the writing of the HAMR, the recording media needs to be heated to near its Curie temperature to lower the coercivity. To avoid the thermal destabilization of adjacent tracks, the recording material should possess an appropriate thermal conductivity ($\kappa$) anisotropy, so that the heat flow is preferential in the cross-plane direction. Based on the FePt/C multilayer structure, Hoan et al [4] reported that the cross-plane $\kappa$ of FePt film is 4.6 ± 4.1 times higher than the in-plane $\kappa$ of FePt–C granular system with 14.5 nm grain size, which is estimated as 1.8 ± 0.2 Wm⁻¹ K⁻¹. However, this result was obtained on the non-crystallized FePt film. In the MgO substrate/FePt(100 nm) system, Ashutosh et al [5] measured the room temperature cross-plane $\kappa$ of FePt films in L1₀ and A1 phase, which were 11.5 ± 0.8 and 8.8 ± 0.6 Wm⁻¹ K⁻¹, respectively. By the heat Fourier equation, Sumei et al [6] reported that the $\kappa$ in the grain boundary can also strongly

Abstract

The structure, magnetic and thermal properties of (FePt₀.78–C₀.22)–BN granular films with various BN volume concentrations are systematically studied. Compared to the FePt–C film without BN doping, the grain size distribution of the FePt grains is reduced by 50% with BN component is 10 vol.%, and the coercivity and the intrinsic switching field distributions of the film are increased 3 times and narrowed by 35%, respectively, when BN component is 6 vol.%. The improvement in magnetic properties is attributed to the enhanced chemical ordering of the FePt L1₀ phase. Based on the 3ω method, it is found that the relative cross-plane thermal conductivity $\kappa_{f, \perp}$ of the FePt–C–BN layer is more correlated to the grain size distribution of the film, which decreases to the minimum value at 10 vol.% BN. Our results indicate that a certain amount of BN doping can improve the comprehensive performance of the FePt–C–BN granular films, making it a promising candidate for heat assisted magnetic recording.

Keywords: FePt, switching field distribution, Thermal conductivity, BN, heat assisted magnetic recording (HAMR)
influence the in-plane and cross-plane $\kappa$ of FePt granular film. Since the grain boundary material also affects the chemical ordering of the FePt film, [7] it is necessary to systematically study the relationship between the grain boundary material and the performance of the FePt granular film.

In order to isolate nano-sized L10 FePt grains, one common method is to introduce non-ferromagnetic material as segregation. In FePt–X film, we first reported that the carbon(C) could enhance the coercivity as well as reduce the size of FePt grains [8]. Subsequently, Steven et al [9] reported that the boron(B) could promote the columnar growth, and C produced the least reduction of chemical ordering of the FePt grains. More recent, our results further indicated that the grain size distribution of the FePt grains can be effectively decreased and unified in the FePt–C–X(SiO2, SiN, ZrO2) system [10–12]. In this work, we studied how the structure, magnetic and thermal properties of the (FePt–C)–BN granular film can be affected by different BN components. Compared to the FePt–C film without BN doping, we found that the coercivity and the intrinsic switching field distribution (SFD) of the FePt–C–BN film are enhanced by 3 times and reduced by 35%, respectively, when the volume concentration of BN is increased to 6 vol.%. At the same time, both the relative cross-plane thermal conductivity and the average grain size distribution of the FePt–C–BN film are reduced to a minimum at 10 vol.% BN. By correlating the characteristic changes, we evaluated the feasibility of the improved FePt–C–BN granular film in HAMR applications.

**Experiment**

The film structure of the sample is shown in figure 1(a). In the thin film depositing, the L10 ordered (FePt0.78C0.22)–BN films with the various volume concentrations of BN doping were deposited on MgO (0 0 1) single crystal substrate by a magnetron sputtering system. The base pressure is lower than $2 \times 10^{-8}$ Torr. By co-sputtering the FePt, C and BN targets, the volume concentration of BN in the samples are modified as 0, 2.7, 6, 7.5, 10 and 15 vol.%, respectively. The substrate temperature is set as 480 °C and the thickness of the (FePt–C)–BN layer is fixed to 4 nm to keep it a good granular structure. After the sample cools down to room temperature, 10 nm thick of SiO2 layer was grown as the insulating capping layer. To measure the thermal conductivity $\kappa$ of the MgO (001) substrate and the (FePt–C)–BN layer, a Ti(3 nm)/Au(50 nm) heater line, which serves as both the heat source and the thermometer, was fabricated on the top of the films by standard lithography process. The size of the heater line is $l = 1$ mm and $2b = 40$ μm, as shown in figure 1(b).

The magnetic hysteresis loops were measured by a superconducting quantum interference device (SQUID) system. The x-ray diffraction (XRD) spectrum was used to analyze the structure and the chemical ordering degree of the films. The average grain diameter distributions were analyzed based on the scanning electron microscopy (SEM). The thermal conductivity $\kappa$ is derived by a self-build system based on the 3ω method [13, 14], as shown in figure 1(c). In the measurement, we utilized a lock-in amplifier to output an alternating AC current at frequency $\omega$ ($=2\pi f$) and measure the resulted third harmonic component ($V_{3\omega}$) of the voltage of the heater line. The sample was measured in a vacuumed environment to reduce the heat loss by thermal radiation. In the heater line, the AC current caused temperature increase $\Delta T_{AC}$ is obtained by the relation [15]:

$$\Delta T_{AC,\text{real}} = \frac{2V_{3\omega,\text{real}}}{\alpha V_{1\omega,\text{real}}}, \quad (1)$$

where $\alpha$ is the temperature coefficient of the heater, and $V_{3\omega,\text{real}}$ ($V_{1\omega,\text{real}}$) is the real part of the $V_{3\omega}$ ($V_{1\omega}$) of the heater line. The temperature coefficient $\alpha$ of the heater line is obtained by fitting the resistance $R(T)$ versus temperature ($T$) curve with the relation:

$$\alpha = \frac{\Delta R(T)}{\Delta T \Delta R_0}, \quad (2)$$

where $R_0$ is resistance of the heater line at 0 K. Based on the 2D heater line model [13–15] and by considering the heat flux out of the heater is equal to the average power ($P$) dissipated by the heater line, the solutions to the heat equation indicate that the temperature increase in the substrate is $\Delta T_s(\omega) = \frac{P}{\pi \kappa_s} \int_0^\infty \frac{\sin^2(bk)}{(bk)^2 + \omega^2/D} dk$, where $\kappa_s$ and $D$ is the thermal conductivity and thermal diffusivity of the substrate. In the limit of $b \sqrt{D/\omega} \ll 1$, the solution to the equation is [16]

$$\Delta T_s(\omega) = \frac{P}{\pi \kappa_s} \left[ \frac{1}{2} \ln \omega - \frac{1}{2} \ln \left( \frac{ib^2}{D} \right) + \text{constant} \right]. \quad (3)$$

Combine the equations (1) and (3), the $\kappa_s$ of the substrate is obtained as

$$\kappa_s = -\frac{\alpha \Delta V_{1\omega,\text{real}}}{4\pi l \kappa_f} \frac{d}{dV_{3\omega,\text{real}}}. \quad (4)$$

When a thin film layer with thickness of $d_f$ ($\ll 2b$) is deposited on the substrate, the 2D heater line model indicates that the film causes a frequency-independent increase to the temperature of the heater line [15]. If the thermal conductivity of the film layer is smaller than the that of the substrate ($\kappa_f < \kappa_s$), the AC current induced temperature increase is:

$$\Delta T_f(\omega) = \Delta T_s(\omega) + \Delta T_f = \Delta T_s(\omega) + \frac{P d_f}{2b \kappa_{f,\perp}}. \quad (5)$$

where $\Delta T_f$ is the temperature increase induced by the film. Corresponding to equation (1), the cross-plane thermal conductivity of the interest film $\kappa_{f,\perp}$ can be calculated by the average temperature rise difference, which is experimentally measured at the same power input by similar heater line deposited on the sample and substrate (or a reference sample) [17]:

$$\kappa_{f,\perp} = \frac{d_f}{2b l} \left( \frac{\Delta T_f}{P} \right)_{s+f}^{-1} = \left( \frac{\Delta T_f}{P} \right)_{s,\text{average}}^{-1}. \quad (6)$$

To obtain the in-plane thermal conductivity of the film layer $\kappa_{f,\parallel}$, one method is to deposit the film on a substrate which has a much smaller $\kappa_s$, which results in the generated heat
Figure 1. (a) The cross-plane schematic of the film structure and the heater line. (b) The top view schematic of the heater line. (c) The schematic of thermal conductivity measuring system.

Figure 2. (a)–(f) The grain size distribution of the (FePt−C)−BN samples with 15, 10, 7.5, 6, 2.7 and 0 vol.% BN, respectively. (g) to (l) The planer view SEM results of all (FePt−C)−BN samples. The corresponding inserts show the counts of the grain diameter.
spreads along the film layer before it enters the substrate.[18] Therefore the temperature increase in the substrate in equation (5) can be neglected.

Results and discussion

In figures 2(g)–(l), we show the planer view SEM images of our samples with 0, 2.7, 6, 7.5, 10 and 15 vol.% BN, respectively. Combining our previous results based on L10 ordered FePt–C and FePt–C–SiNx films, [8, 10] it is indicated that the (FePt–C)–BN film layer in all samples have well-isolated nanoscale granular structure. In figures 2(a)–(f), we show the corresponding grain size distribution of these samples. The average grain size distribution rapidly decreases from 28 ± 8 to 19 ± 7 nm when BN component is increased from 0 to 2.7 vol.%. As the BN doping is further increased, the average grain size distribution of the sample gradually decreases to a minimum value of 14 ± 2 nm with 10 vol.% BN. Since the C atoms prefer to diffuse to the film surface [10], the narrower grain sized distribution implies that the BN atoms diffuse more at grain boundaries. As the BN component is further increased to 15 vol.%, the average grain size distribution of the film is increased to 15 ± 3 nm. This may be due to the diffusion of over doped BN between FePt and substrate, which hinders the epitaxial columnar growth of FePt grains [19].

Figures 3(a)–(f) show the normalized in-plane (IP) and out-of-plane (OP) magnetic hysteresis (M–H) loops of the samples with BN components varying from 0 to 15 vol.%. The saturate magnetization \( M_s \) of all samples is measured around 1000 ± 50 emu/cc, which is comparable with the reported \( M_s \) of L10-ordered FePt film.[7] In the M–H loops, it can be seen that the remanence in all IP loops is close to zero and OP loops is close to 100%. The magnetization of the samples with 2.7 to 10 vol.% BN cannot be saturated at the magnetic field of 75 kOe. This means that the perpendicular magnetic anisotropy energy \( K_u \) of these samples are larger than 8.1 \( \times \) 10^7 erg/cc. We extract the coercivity \( H_c \) value of all OP loops and show them in figure 3(h). It can be seen that the \( H_c \) increases by 3 times, which from 6 kOe to 18 kOe when BN component is increased from 0 to 6 vol.%. But it starts to decrease when the BN doping is further increased. This variation is inconsistent with the results reported by Danniil et al [20], indicating that the smaller grain size in FePt/BN system provides less coercivity. On the other hand, it was reported that the chemical ordering could provide different influences on the thin film coercivity [21–23].

To understand the variation of \( H_c \) on the BN components, we measured the XRD patterns of all samples and show them in figure 3(g). It can be seen that only the MgO(002) peak, the FePt(001) peak and FePt(002) + (200) peak are found, indicating the C and BN are in amorphous state. When there is no BN doping, the out of plane direction lattice constant \( c \) of the sample is 0.377 nm, which is larger than that of the bulk L10 bulk FePt (0.373 nm). This is because the 22% C
concentration separates the FePt into granular structure, as shown in figure 2(k), which leads to the reduction of the strain acting on the FePt layer and the decrease of the chemical ordering [11]. As the BN component is increased from 0 to 6 vol.%, the FePt(0 0 1) peak shifts to a large angle and the lattice constant $c$ is decreased to 0.375, indicating an improvement in the chemical ordering. When the BN doping is further increased, the FePt(0 0 1) peak shifts to a low angle and the lattice constant $c$ linearly increase to 0.380. The movement of the FePt(0 0 2) $+$(2 0 0) peak also reveals a more pronounced and similar trend. When BN component is increased to 6 vol.%, the FePt (0 0 2) $+$(2 0 0) peak shifts more to FePt (0 0 2) peak position. Whereas when BN component is further increased to 15 vol.%, it shifts to FePt (2 0 0) peak position. We show the dependence of the chemical ordering of all samples, which is estimated by the square root of the integrated peak intensity ratio $(I_{001}/I_{002})^{1/2}$, on the BN components in figure 3(h). It can be seen that the $(I_{001}/I_{002})^{1/2}$ value reaches the maximum value of 0.55 with 6 vol.% BN, which shows a similar trend as the $H_c$. At the same time, the lattice constant ration $c/a$ for all samples are calculated around 0.98–0.99, indicating the stress/strain is not the main reason to lead the $H_c$ variation. As a result, the increase of the $H_c$ in our sample is mainly attributed to the chemical ordering improvement of the FePt L10 phase, and the reduction of the $H_c$ is due to the existent of soft fcc FePt phase in small grain size, which is exchange coupled with Li$_3$FePt grains [20, 22–24].

As the ferromagnetic layer with a narrow intrinsic SFD is preferred in the HAMR application, the effect of BN doping on the SFD performance of our samples are further evaluated. Based on the $\Delta H_i/H_c$ method reported by Berger et al [25, 26], we show the process of obtaining the intrinsic SFD of the sample with 6 vol.% BN in figures 4(a) and (b). After normalizing and transposing the original data from the main and recoil $M$–$H$ loops as shown in the inset of figure 4(a), we obtain the normalized field difference $\Delta H_i/H_c$ between the partial reversal curve and major loops. Then the complete data set of $\Delta H_i/H_c$ versus $M/M_s$ curves are fitted with the relation [25, 26]:

$$\Delta H_i(M, \Delta M) = \frac{1}{2} \left[ \frac{1 - M}{\sigma^2} \right] - \frac{1}{2} \left[ \frac{1 - (M + \Delta M)}{\sigma^2} \right],$$

$$\frac{1}{2} \left[ \frac{1 - M}{\sigma^2} \right] = -\sqrt{2} \Delta \sigma \Delta \text{erf}^{-1}(M) - \omega \tan(\frac{\pi}{2} M) - \frac{1}{1 + \beta M},$$

where $\sigma$, $\alpha$, $\omega$ and $\beta$ are the fitting parameters. In figure 4(b), it can be seen that the fit quality is good, of which the high

![Figure 4](image-url)
multiple correlation coefficients $R^2$ is 0.997. The intrinsic SFD $D(H_s)$ of the sample is obtained by substituting the parameters into the following relation [25, 26]:

$$H_s(M) = -I^{-1} \left[ 1 - \frac{M}{2} \right],$$  \hspace{1cm} (9)$$

$$H_s(M) \rightarrow M(H_s),$$  \hspace{1cm} (10)$$

$$\frac{dM(H_s)}{dH_s} = D(H_s).$$  \hspace{1cm} (11)$$

In figures 4(c)–(f), we show the field-integrated intrinsic SFD of the samples with 0, 2.7, 6 and 7.5 vol.% BN, respectively. By the Gaussian equation:

$$D_G(H_s) = \frac{1}{\sqrt{2\pi}\sigma_G} \exp \left[ -\frac{2(H_s - H_G')}{2\sigma_G} \right],$$  \hspace{1cm} (12)$$

the normalized SFD value $\sigma_G/H_G'$ of the corresponding sample is labeled in figures 4(c)–(f). When BN component is increased from 0 to 6 vol.%, the SFD of the sample reduces by 35%, which from 35.9% to 23.3%. While when BN component is further increased to 7.5 vol.%, the SFD of the sample increases to 35.1%. Combine the results from figure 3(h), it can be suggested that the SFD variation of our samples is partly derived from the chemical ordering changes. It should be noted that the $\Delta H(M, \Delta M)$ method is based on the non-thermally interacting hysteron model with exchange and dipolar interactions, and it does not consider the reversible magnetization components [27]. As a result, the SFD of the samples with 10 and 15 vol.% BN cannot be properly measured by this method because they possess more exchange pairs between FePt L10 phase and the soft fcc phase.

To test the setup of our thermal conductivity measurement system, we first measure the $\kappa$ of a pure MgO (0 0 1) substrate. As shown in figure 5(a), we linearly fit the $R$–$T$ curve of the heater line and obtain its slope value, which is 0.03 $\Omega$ K$^{-1}$. By linear fitting the $V_{3\omega}$, real versus $\ln \omega$ curve, as shown in figure 5(b), and based on equation (4), the thermal conductivity of the MgO (0 0 1) substrate is obtained as $\kappa_{MgO} = 65.4$ Wm$^{-1}$ K$^{-1}$. This result is comparable with the value $\kappa_{MgO} \sim 60.1 \pm 5.8$ Wm$^{-1}$ K$^{-1}$ measured based on the time domain thermoreflectance (TDTR) technique [5].

In figure 6(a), we show the dependence of the temperature increase $\Delta T$ on the measuring frequency $\ln \omega$ of our thin-film samples with 0, 2.7, 6, 7.5, 10 and 15 vol.% BN, respectively. Based on equation (5), it can be known that the $\Delta T$ in our thin-film samples is contributed by two parts: one is the relative MgO substrate and the other one is the film layers. Since our samples possess the same layer thickness, if the temperature increases contributed by the MgO substrate and the SiO$_2$ layer are identical, it can be indicated that the relative $\Delta T_f$ offset, as indicated in figure 6(a), is mainly derived from the FePt–C–BN film layer. We extracted the relative $\Delta T_f$ offset value and show its reciprocal dependence on BN component in figure 6(b). The positive $(\Delta T_f)^{-1}$ value indicates the $\kappa_{f,\perp}$.
of all FePt–C–BN films with BN doping is smaller than that of the FePt–C film. At the same time, the relative \((\Delta T_f)^{-1}\) value decreases to minimum with 10 vol.% BN, which has a similar trend with the grain size distribution, as shown in figure 6(b). The results from Ashutosh et al [5], show that the chemical ordering enhancement of the continuous FePt film can lead to the increase of film cross-plane thermal conductivity. In our cases, it can be found that the cross-plane thermal conductivity of the FePt–C–BN granular film is more related to its grain size distribution state. Although doping BN in FePt–C film leads to the reduction of \(\kappa_{f,\perp}\), it is known that the reduced grain size also causes the in-plane thermal conductivity \(\kappa_{f,||}\) to decrease, due to the presence of more FePt/(C–BN) interfaces per unit length [4]. Therefore the thermal conductivity ratio of \(\kappa_{f,\perp}/\kappa_{f,||}\), which is more important in the HAMR applications, will show a smoother decline or even an upward trend with the appearance of BN doping.

**Conclusion**

We systemically studied the structure, magnetic and thermal properties of (FePt–C)–BN granular film with various volume concentrations of BN doping. The magnetic properties of the FePt–C–BN film show the maximum coercivity of 18 kOe and minimum SFD of 23.3% are obtained when BN component is 6 vol.%. The improvement in magnetic properties is mainly attributed to the chemical ordering increase of FePt L10 phase. Compared with the change of magnetic properties, we found that the relative cross-plane thermal conductivity of FePt–C–BN film is more related to its grain size distribution state. Both the average grain size distribution and the relative \(\kappa_{f,\perp}\) of the FePt–C–BN film decrease to the minimum when BN component is 10 vol.% By balancing the structure, magnetic and thermal properties through doping a certain amount of BN, we believe the FePt–C–BN granular film can be a promising candidate for the recording material in the HAMR applications.

**Acknowledgment**

This work is partially supported by Seagate Technology, Singapore Ministry of Education MOE2018-T2-2-043, AMEIRG18-0022, A*STAR IAF-ICP 11801E0036.

**ORCID iDs**

Jingsheng Chen https://orcid.org/0000-0003-3188-2803

**References**

[1] Weller D, Mosendz O, Parker G, Pisana S and Santos T S 2013 *Phys. Status Solidi a* 210 1245–60
[2] Zhang L, Takahashi Y K, Hono K, Stipe B C, Juang J and Grobis M 2011 *IEEE Trans. Magn.* 47 4062–5
[3] Bain J A, Malen J A, Jeong M and Ganapathy T 2018 *MRS Bull.* 43 112–8
[4] Ho H, Sharma A A, Ong W L, Malen J A, Bain J A and Zhu J G 2013 *Appl. Phys. Lett.* 103 131907
[5] Giri A, Wee S, Jain S, Hellwig O and Hopkins P E 2016 *Sci. Rep.* 6 32077
[6] Wang S and Victora R H 2015 *Appl. Phys. Lett.* 117 17D147
[7] Lyubina J, Rellininghaus B, Gutleisch O and Albrecht M 2011 *Handbook of Magnetic Materials* 19 291–407
[8] Chen J S, Lim B C, Hu J F, Liu B, Chow G M and Ju G 2007 *Appl. Phys. Lett.* 91 132506
[9] Granz S D, Barmak K and Kryder M H 2013 *Eur. Phys. J. B* 86 81
[10] Dong K F, Li H H, Peng Y G, Ju G, Chow G M and Chen J S 2012 *J. Appl. Phys.* 111 07A308
[11] Li H, Dong K, Peng Y, Ju G, Chow G M and Chen J S 2013 *Appl. Phys. Express* 6 075502
[12] Dong K F, Li H H, Deng J Y, Peng Y G, Ju G, Chow G M and Chen J S 2015 *J. Appl. Phys.* 117 17D116
[13] Cahill D G, Fischer H E, Klitsner T, Swartz E T and Pohl R O 1989 *J. Vac. Sci. Technol. A* 7 1259–66
[14] Kim J H, Feldman A and Novotny D 1999 *J. Appl. Phys.* 86 3959–63
[15] Zhao D, Qian X, Gu X, Jajja S A and Yang R 2016 *J. Electron. Packag.* 138 040802
[16] Kimling J, Nielsch K, Rott K and Reiss G 2013 *Phys. Rev.* B 87 134406
[17] Borca-Tasciuc T, Kumar A R and Chen G 2001 *Rev. Sci. Instru.* 72 2139–47
[18] Kimling J, Wilson R B, Rott K, Kimling J, Reiss G and Cahill D G 2015 *Phys. Rev. B* 91 144405
[19] Deng J, Li H, Dong K, Li R W, Peng Y, Ju G, Hu J, Chow G M and Chen J 2018 *Phys. Rev. Appl.* 9 034023
[20] Daniil M, Farber P A, Okumura H, Hadjipanayis G C and Weller D 2002 *J. Magn. Magn. Mater.* 246 297–302
[21] Luo L, Anuniwat N, Dao N, Cui Y, Walf S A and Lu J 2016 *J. Appl. Phys.* 115 054422
[22] Christodoulides J A, Farber P, Daniil M, Okumura H, Hadjipanayis G C, Skumryev V, Simopoulos A and Weller D 2001 *IEEE Trans. Magn.* 37 1292
[23] Wang J P, Shen W K, Bai J M, Victora R H, Judy J H and Song W L 2005 *Appl. Phys. Lett.* 86 142504
[24] Berger A, Lengsfield B and Ikeda Y 2006 *J. Magn. Magn. Mater.* 300 1249–52
[25] Bain J A, Malen J A, Jeong M and Ganapathy T 2018 *MRS Bull.* 43 112–8
[26] Liu Y, Dahnem K A and Berger A 2003 *J. Appl. Phys.* 93 9902
[27] Christodoulides J A, Farber P, Daniil M, Okumura H, Hadjipanayis G C, Skumryev V, Simopoulos A and Weller D 2001 *IEEE Trans. Magn.* 37 1292