Magnetic properties of $L1_0$-FePt/Fe exchange-coupled composite nanodots

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The magnetic properties of the $L1_0$-FePt/Fe exchange-coupled composite nanodots were investigated by Hall effect measurement due to its high sensitivity. The FePt/Fe nanodots showed that the coercivity changed irregularly with respect to the Fe thickness, which was deviated from the simulated results where the coercivity of the ECC dots reduced with the increase of Fe soft layer thickness. It was found that the edge damage induced by ion-milling affected the coercivity of the nanodots significantly and the magnetization reversal mechanism. Domain wall nucleation and propagation was revealed in the FePt single domain dots. In the FePt/Fe ECC nanodots, the magnetization gradually reversed from the out-of-plane direction towards the in-plane direction before switching to the reversed out-of-plane direction. The critical size of magnetic single domain in the FePt/Fe ECC nanodots increased with the increasing of Fe soft layer thickness.

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1. Introduction

The development of nanotechnology in the last three decades has made the things become smaller and work more efficiently as foreseen by Richard Feynman in his famous talk “There is plenty of room at the bottom” in 1959 [1]. Nanostructured materials are widely used in information storage, electronics, bioengineering, medical application, optical engineering, energy conservation, environmental protection, nanofabrics, cosmetics, defence & security etc. The innovation of applying magnetic materials into the information storage technology, such as the magnetic recording thin film, the giant magnetoresistance (GMR) sensor [2], and the tunneling magnetoresistance (TMR) sensor [3] used in the hard disk drives, provides us the cheapest data storage today at the cost of few cents for storing 1 Gbyte of information. The emerging non-volatile magnetic random access memories (MRAMs) [4] are believed to boost the recording speed further. With the continuous reduction in feature size, magnetic materials of large magnetic anisotropy are needed to sustain the thermal stability for a storage life of more than 10 years. One of the most potential candidates for the future recording technology, such as patterned magnetic recording media [5], heat-assisted magnetic recording (HAMR) media [6], and spintronics [7], is the $L1_0$-FePt which has an extremely high magnetoanisotropy ($7 \times 10^7$ erg/cc) and good correction resistance [8]. However, writability is a concern for using high anisotropic magnetic materials. The switching field of $L1_0$-FePt is far beyond the available writing field. To overcome this problem, exchange-coupled composite (ECC) [9] was introduced to assist the magnetization switching by exchange-coupling the magnetically soft phase to the magnetically hard phase. Bilayer [10], ledge-type [11], and core-shell [12] structured $L1_0$-FePt ECC have been widely reported. However, the shape deviates from the proposed ECC structure where the soft and hard phases stacking exactly on the other. This deviation may affect the magnetization reversal mechanism of the $L1_0$-FePt ECC. To achieve the exact ECC structure, $L1_0$-FePt/Fe ECC nanodots with the Fe soft phase stacking on top of the FePt hard phase was fabricated by
patterning the $L1_0$-FePt/Fe bilayer. In general, there are two types of approaches to create the nanopatterns: top-down processes which are based on direct lithography and nanoimprint, and bottom-up processes which use polymer self-assembly [13]. $L1_0$-FePt nanodots have been fabricated by electron beam lithography (EBL) [14-15], nanoimprint lithography [16-17], and copolymer self-assembly [18], respectively, followed by dry etching. In our work, the nanodots were fabricated by the top-down method involving EBL and ion milling.

As the feature size reduces, restudying the magnetic properties including magnetization reversal and magnetic domain structure is needed since the magnetic properties are size-dependent. However, characterizing the magnetic properties of the $L1_0$-FePt nanodots is another challenge because the commercial equipment such as vibrating sample magnetometer (VSM), alternating gradient force magnetometer (AGFM), and superconducting quantum interference device (SQUID) detection system has a sensitivity limit of $10^{-6}$ emu which requires large patterned area to provide enough signals. Patterning in a large area is not effective for research as it encounters the uniformity problem during nanoimprint or copolymer self-assembly, and the cost and time problem when use EBL. The traditional way is to use magnetic force magnetometer (MFM) to measure the switching status of the dot array [14, 19-21]. However, there are disadvantages of this characterization method: the quality of the MFM images is affected by the MFM tip resolution especially for dot size smaller than 20 nm; only remanent magnetization can be measured; there is no temperature variation; angular dependent measurement is invalid; and the dot-counting is inconvenient. An alternative way is to use the highly-sensitive anomalous Hall effect (AHE) measurement [22-24] to characterize the magnetic properties of the magnetic nanodots with perpendicular magnetization by composing the nanopatterns into the Hall bar. It can provide more types of measurements such as angular- and temperature-dependent hysteresis loops and magnetic recoil loops.

In the current work, the patterned $L1_0$-FePt nanodots and $L1_0$-FePt/Fe ECC nanodots were used to simulate the exact ECC structure which consists of one soft phase and one hard phase
stacking in the same columnar grain. Highly-sensitive AHE measurements were applied to investigate the magnetic properties and reversal mechanism of the nanodots. Since the patterning processes involved ion milling which would result in extrinsic magnetic defects that modify the magnetic properties and domain structure of the magnetic film as found in the (Co/Pd)$_n$ multilayer nanodots [25-26], it is crucial to understand the effect of the edge damage on the magnetic properties and magnetization reversal mechanism. Although some research groups have investigated the magnetic properties of the $L_1\text{I}_0$-FePt based ECC nanopatterns [27-29], a detailed study on the effect of ion milling damage and the Fe film thickness on the magnetic properties of the $L_1\text{I}_0$-FePt/Fe ECC nanodots has not been reported. Here a systematic study of the effect of Fe film thickness and ion milling damage on the magnetic properties, magnetization reversal and magnetic domain structure was reported.

The manuscript was organized as follows. A detailed description of the sample fabrication, the measurement setup and the micromagnetic simulations are given in Section 2. In Section 3 the coercivity, the magnetization reversal mechanism and magnetic domain structure of the $L_1\text{I}_0$-FePt and $L_1\text{I}_0$-FePt/Fe ECC nanodots are presented and discussed. The effect of the ion-milling damage on the magnetic properties was investigated. A short conclusion is given in Section 4.

2. Experiments and methods

The nanopatterns were fabricated by the deposition of continuous magnetic films followed by EBL and Ar$^+$ milling. 16 nm $L\text{I}_0$-FePt film was sputter-deposited on the MgO (100) single crystal substrates at 400°C, followed by in-situ annealing at 450°C for 30 min to enhance the chemical ordering and film continuity. The target composition is Fe$_{55}$Pt$_{45}$. Fe thin layer with various thicknesses (0, 2, 4, 6 and 8 nm) was deposited directly on the FePt layer after the sample cooled down to room temperature. A thin Au over layer was deposited to protect the Fe layer from oxidation.
The patterns were transferred onto the 50-nm maN2401 resist by Elionix ELS-7700 EBL system with 100kV acceleration voltage. Ion milling was carried out in Veeco RF-350 Ion-Beam Milling System with the pre-calibrated etching rate. The dot array has an area of 20×20 µm. The dot diameter varies from 50 nm, 100 nm, 250 nm, to 500 nm. The space among the adjacent dots is fixed at 100 nm. After the ion milling, a 24-28 nm Ta was deposited to fill up the gap among the dots to facilitate the electron transportation during Hall measurement. Because FePt and Fe have higher conductivity than Ta, the electric current can pass through the dot array. The Hall devices with cross Hall bars were fabricated on the dots and Ta film by a contact mask aligner (SUSS MicroTec MA8) using 365 nm UV light for microscale patterns.

The magnetic properties of the nanodots were investigated by Hall effect measurement. The setup for Hall voltage measurement is shown in figure 1(a). The measurement was carried out in the Physical Property Measurement System (PPMS) which can supply a maximum magnetic field up to 9 Tesla.

The Hall voltage has contributions from the normal Hall voltage which is present for all conductive materials, the anomalous Hall voltage (AHV) which depends on the out-of-plane magnetization of the ferromagnetic material, the planar Hall voltage (PHV) which depends on the in-plane magnetization of the ferromagnetic material, and the voltage off-set caused by misalignment. A schematic of the measurement is shown in figure 1(b). The Hall voltage (V_H) can be expressed by the following equation [30-31]:

\[
V_H = \frac{R_0 I}{t} H \cos \varphi + \frac{\mu_0 R_s I}{t} M \cos \theta + \frac{kI}{t} M^2 \sin^2 \theta \sin 2\phi + V_{offset}
\]  

(1)

where \( I \) is current, \( t \) is layer thickness, \( M \) is magnetization, \( V_{offset} \) is off-set voltage, \( R_0 \) is normal Hall coefficient, \( R_s \) is anomalous Hall coefficient, and \( k \) is a constant related to the anisotropic magnetoresistance effect. \( R_s \) is usually one or two orders higher than the other two coefficients.

The out-of-plane magnetization and in-plane magnetization at each magnetic field can be
extracted from the measurement. The detailed data extracting processes can be referred to Ref. 20 and 26.

![Figure 1](image)

**Figure 1.** The experimental setup (a) and schematics (b) of Hall voltage measurement. The dot arrays are centered in the cross of the Hall bar. The magnetic field is applied perpendicular to the sample surface, and the electric current is applied in the longitudinal direction, while the Hall voltage is measured in the transverse direction perpendicular to both the magnetic field and the current.

Micromagnetic simulations were carried out for insight understanding. Here the commercial software - LLG Micromagnetic Simulator, was used [32]. The magnetic parameters used for simulation came from the experimental measurements. The magnetic anisotropy of the $L1_0$-FePt magnetically hard layer was calculated by the generalized Sucksmith and Thompson (GST) method [33-35] which analyzed the normalized magnetization curve measured by AHE method. The linear fitting gave $K_f^{\text{eff}}$: $1.45 \pm 0.02 \times 10^7$ erg/cc, and $K_2$: $3.43 \pm 0.03 \times 10^6$ erg/cc. Taking into account the demagnetization energy [36], the final anisotropy $K_f$ is $2.08 \pm 0.02 \times 10^7$ erg/cc. The calculated result agrees quite well with the value calculated from the enclosed area between the easy-axis magnetization curve and the hard-axis magnetization curve. During simulation, the FePt hard phase has the following magnetic parameters: $M_s^h = 1000$ emu/cc, $K_{u1}^h = 2.0 \times 10^7$ erg/cc, $A^h = 1.05 \times 10^6$ erg/cm, $t^h = 16$ nm, it has the uniaxial anisotropy with easy axis along the z-direction; the Fe soft phase with a cubic anisotropy has the following magnetic parameters: $M_s^s = 1716$
emu/cc, \( K' = 4.7 \times 10^5 \text{ erg/cc}, A' = 2.10 \times 10^{-6} \text{ erg/cm}, \tau' = 4 \text{ nm} \). The interlayer exchange stiffness is assumed to be \( 1.0 \times 10^{-6} \text{ erg/cm} \). The mesh size used is \( 1 \times 1 \times 1 \text{ nm} \). The simulation was carried out at 0 K using the convergence \( (1.0 \times 10^{-8}) \), which is the absolute value of the change during the iteration process of any direction cosine component, as the exit criteria for the calculation.

3. Results and discussion

3.1. Magnetic properties of \( L1_0 \)-FePt and \( L1_0 \)-FePt /Fe ECC nanodots

In the XRD pattern, only (001), (002) and (003) peaks of the \( L1_0 \) phase were present in the FePt continuous film, indicating the formation of perpendicular magnetization. The rocking curve of (001) peak showed sharp full-width at half-maximum (FWHM), implying a narrow easy axis distribution. X-ray Reflection (XRR) measurement had confirmed that the FePt layer had a thickness of 16 nm and a surface roughness around 3 nm. After the deposition of Fe magnetically soft layer, the (001) texture of the \( L1_0 \)-FePt magnetically hard phase was maintained. In contrast to the Fe bcc (110) texture formed in the FePt/Fe exchange-coupled composite reported by Ma et al. [37], bcc (200) texture was observed in the thicker Fe layer of the FePt/Fe exchange-coupled bilayer in the XRD diffraction pattern. The high-resolution TEM image (not shown here) further proved that Fe layer grew epitaxially on the (001)-textured FePt layer.

Figure 2 compares the out-of-plane M-H curves before and after patterning for the FePt single layer and the FePt/Fe (4 nm) bilayer, respectively. The large squareness and remanence in figure 2(a) further confirms that the easy axis of FePt layer is in the out-of-plane direction, and the magnetization is reversed by domain-wall motion as the lateral exchange-coupling is strong due to the film continuity. After patterning the coercivity of the nanodots is 4-5 times that before patterning. This large increase in coercivity after patterning is due to the elimination of long-range lateral exchange-coupling among the dots by physical separation and thus preventing domain wall motion during magnetization reversal. This was demonstrated by the MFM images.
in figure 9 which showed the transition of magnetic multidomains to single domain after patterning. The normalized remanent magnetization is sensitive to the demagnetization. Patterning increased the remanence of the bilayer. Before patterning, the bilayer had a demagnetization factor $N_z$ around 1. After patterning $N_z$ was reduced to 0.66 for the 100 nm dots and 0.51 for the 50 nm dots. Hence, without any external magnetic field, the bilayer film experienced a larger reversed demagnetization field than the nanodots did.

![Normalized out-of-plane hysteresis loop](image)

**Figure 2.** Normalized out-of-plane hysteresis loop of the FePt single layer (a) and FePt/Fe(4 nm) bilayer (b) before patterning and after patterning.

The out-of-plane coercivities of the FePt/Fe ECC bilayers before and after patterning are summarized in figure 3. Although the coercivity of FePt layer increased dramatically after patterning, it (9.4 - 9.8 kOe) is still far below the anisotropy field - 36.1 kOe according to simulation. Similar phenomenon has been reported by other groups [22, 38]. One reason may be due to the intrinsic defects inside the FePt film which cannot be avoided during sputtering. The other reason may be due to the magnetic inhomogeneity at the dot edge which was damaged by ion milling. Before patterning, the coercivity of the continuous bilayer reduced with the increase of Fe soft layer thickness as shown in our experimental results. However, it showed the discrepancy of the coercivity with Fe thickness in the nanodots after patterning comparing to the simulation results. The micromagnetic simulation showed that the coercivity of the nanodots
reduced with the increase of Fe layer thickness. The discrepancy of the coercivities with Fe thickness after patterning might be largely due to the deterioration of magnetic properties at the damaged edge. The effect of ion milling damage on the magnetic properties of the FePt and FePt/Fe ECC nanodots will be discussed below.

**Figure 3.** Top: coercivity of the FePt/Fe ECC bilayer before and after patterning based on experimental results; Bottom: simulated coercivity of the FePt/Fe ECC dots with various Fe thicknesses.

The soft phase inside the FePt layer was revealed in the magnetic recoil loops shown in figure 4(a). In the 100 nm dots, the magnetization was almost unchanged upon removing the reversal magnetic field, indicating there is no soft phase or the magnetically soft phase and hard phase are strongly coupled. In the 50 nm dots, upon removing the magnetic field, a small amount of magnetization reversed the direction, which showed the exchange spring effect. The reversible component was the soft phase which was not fully exchange-coupled to the FePt hard phase. Since the ion milling conditions for the 100 nm dots and the 50 nm dots and the inter-dot gap
were the same, the damage caused by the ion milling for these two types of dots should be similar. Because the 50 nm dots had larger surface-to-volume ratio, the edge damage had larger influence on the magnetic properties of the dots.

![Figure 4](image.png)

**Figure 4.** (a) Recoil loops of the FePt 100 nm and 50 nm nanodots. (b) Left: The slope of the recoil loop is plotted to the corresponding unreversed magnetization for the FePt 100 nm dots; Right: the reversing rate and fitting quality $R^2$ of the FePt 100 nm and 50 nm dots are summarized.

The magnetization reversing rate of the sample was calculated from the slope of the recoil loop. The slope of each recoil loop varies with respect to different amount of unreversed magnetization. When more unreversed magnetization is present, the slope of the magnetization reversing is larger (see recoil loop 1). This is because upon removing the external field the unreversed magnetization forces the reversed soft phase to go back to the unreversed direction through exchange coupling. In contrast, recoil loop 3 has a smaller slope of the magnetization reversing. Here, the slope of the magnetization reversing from each recoil loop was plotted to the corresponding unreversed magnetization as shown in figure 4(b). The slope of the linear fitting gives the reversing rate with the unit of (kOe)$^{-1}$. It was found that the FePt 100 nm dots had a very
small reversing rate of 0.0038 (kOe)$^{-1}$, while the 50 nm dot had a larger reversing rate of 0.0095 (kOe)$^{-1}$. Hence, the 50 nm dot had larger magnetically soft phase-to-hard phase ratio, which agreed with the larger surface-to-volume ratio. This implied that ion milling would damage the dot edge and modify the magnetic properties of the dot.

Micromagnetic simulations were performed to study the influence of the magnetic inhomogeneity at the dot edge on the magnetic properties of the FePt nanodots. It was assumed that the magnetic anisotropy of the dot edge was reduced to 5×10$^5$ erg/cc by the ion milling damage. This assumption was made because the anisotropy of the FePt hard layer mainly relies on its long range $L1_0$ ordering which can be destroyed by Ar$^+$ ion milling. Figure 5(a) showed the simulated coercivity after various damaged edge thicknesses were taken into account. A 2 nm damaged edge reduced the coercivity of the dot by half. Above 6 nm of the dot edge, the coercivity of the 100 nm dot outstood that of the 50 nm dot. Our experimental results showed that the 100 nm dots had larger coercivity than the 50 nm dots, which implied that the thickness of the damaged edge should be large. With 6 nm of damaged edge, the surface-to-volume ratio of the 50 nm dot is 0.4224, which is around 2 times of that of the 100 nm dot. Taking the thermal fluctuation at the surface of the dot into account, the reduction in coercivity of the 50 nm dot would be more compared to the 100 nm dot. Therefore, the damaged edge should not be larger than 6 nm.

The influence of the edge damage on the coercivity of the FePt/Fe nanodot was shown in figure 5(b). A reasonable edge thickness of 4 nm was assumed because the discussion on the FePt nanodots had shown that the edge damage was quite significant. Considering the damaged edge, the coercivity of the FePt/Fe ECC nanodot was less dependent on the Fe thickness compared to the one with no edge damage. If the same amount of edge damage was considered for each FePt/Fe dot, the coercivity still reduced with the increase of Fe layer thickness. However, the experimental results did not show a clear trend. Indeed, the edge damage in each sample differed because the ion milling time increased slightly with the increasing of Fe layer thickness.
Figure 5. (a) The simulated coercivity of FePt 100 nm and 50 nm dots at various damaged edge thickness. (b) Simulated coercivity of the FePt/Fe ECC 50 nm dot without and with 4 nm of edge damage. The inserted schematics are the cross-sectional view of the nanodot with damaged edge.

3.2. Magnetic reversal in the $L1_0$-FePt and $L1_0$-FePt/Fe ECC nanodots

In this section, the magnetization reversal mechanism of the nanodots was characterized by the angular dependent coercivity and the simultaneously measured out-of-plane and in-plane magnetization.

Figure 6(a) shows the angular dependent coercivity of FePt single layer before and after patterning. The coercivity of the continuous film followed the $1/\cos \theta$ curve quite well, which meant the magnetization was reversed by domain wall motion. This is consistent with the magnetization reversal mechanism of the magnetic thin films. It was unexpected that the coercivity of the FePt nanodots increased when the applied field tilted towards the in-plane direction, which showed domain wall propagation instead of coherent rotation in Stoner-Wohlfarth particles. This behavior resembled the magnetization reversal of the exchange-coupled/exchange spring media. Similar phenomenon has been reported in the FePt 60 nm single
The possible explanation is that the magnetization reversal started from the soft phase at the dot edge and propagated into the FePt hard phase.

Figure 6(b) indicates that the magnetization of the FePt/Fe ECC nanodots reversed by domain wall nucleation and propagation. Compared to that of the FePt nanodots, the angular dependent coercivity of the FePt/Fe ECC nanodots became closer to the domain wall motion curve. This is because there is more magnetically soft phase in the FePt/Fe nanodots, which would facilitate the domain wall propagation. However, there was obvious deviation of angular dependent coercivity curve from domain wall motion, which was due to the formation of partial domain wall in the soft phase. The full domain wall formation requires the thickness of the soft layer much larger than the domain wall width of the hard layer, where is $l_{dw} = 4\sqrt{A/K_u} = 8.9nm$ here [39]. However, the maximum Fe thickness in the FePt/Fe ECC nanodots is only 8 nm.

Figure 6. Normalized angular dependent coercivity of FePt single layer before and after patterning (a) and FePt/Fe ECC nanodots with various Fe thicknesses (b).

The Hall measurement provides the relative values of the out-of-plane magnetization $M_\perp$ and the in-plane magnetization $M_\parallel$ under the same applied field, which is very helpful in observing the magnetization reversal processes. Figure 7 shows the plot of anomalous Hall voltage (AHV) and the planar Hall voltage (PHV) to the applied magnetic field for the FePt
nanodots and the FePt/Fe ECC nanodots. The out-of-plane magnetization $M_\perp$ is proportional to the AHV, while the square of in-plane magnetization $M_\parallel^2$ is proportional to the PHV. In the FePt nanodots, the hysteresis loop showed that the magnetization was maintained in the out-of-plane direction before switching. The in-plane magnetization did not change but only at the position where $M_\perp$ reversed its direction. In the FePt/Fe nanodots, the $M_\parallel$ increased but the $M_\perp$ decreased gradually when the magnetic field changed from the positive max to the negative max, the in-plane magnetization reached a peak value where the magnetization switched. The magnetization of the FePt/Fe ECC dots reversed the magnetization in a processional way, with the magnetization switching gradually from the $+z$ direction to the in-plane direction then to the $-z$ direction.
**Figure 7.** The plots of anomalous Hall voltage (AHV) and planar Hall voltage (PHV) to the applied magnetic field for FePt nanodots (a) and FePt/Fe ECC nanodots (b). The range of the AHV is 100 times that of the PHV because the anomalous Hall coefficient is much larger than the planar Hall coefficient. The PHV curve is not centered in the 0V due to the voltage offset which cannot be easily eliminated.

Taking the damaged edge into account, the magnetic domain structure during magnetization reversal of the FePt 50 nm dot and the FePt/Fe (4nm) 50 nm dot was revealed by the micromagnetic simulation as shown in figure 8. In the FePt 50 nm dot, the magnetization switched at one magnetic field -12 kOe. During the switching, the domain nucleated at the edge of the dot and propagated to the center of the dot. This agrees with the reversal mechanism revealed by angular dependent coercivity. In the FePt/Fe(4 nm) ECC 50 nm dot, the magnetization reversal started from the Fe soft phase and the damaged edge, and then propagated to the FePt hard phase, which involved the domain wall nucleation in the soft phase and propagation in the hard phase in both vertical and lateral directions. The switching field of the FePt/Fe ECC dot was around half of that of the FePt hard phase dot. This reduction was due to the assist of Fe soft phase during magnetization reversal through exchange coupling.
Figure 8. The simulated out-of-plane hysteresis loops of the FePt and FePt/Fe(4 nm) 50 nm single dot with 4 nm damaged edge. The cross-sectional magnetic domain structure of the corresponding switching state (red circle) is inserted in the plots. The red color shows the magnetization pointing to the +z direction, and the blue color shows the magnetization pointing to the -z direction.

3.3. Magnetic domain structure

Lots of studies on the magnetic properties and magnetization reversal mechanism of the $L1_0$-FePt based exchange couple/spring bilayer/multilayers have been reported so far. Similar studies on the $L1_0$-FePt based patterned nanodots were less found. However, how the soft layer affects the magnetic domain state of the $L1_0$-FePt-based patterned nanodots has not been reported yet. In this section, the magnetic domain structure of the $L1_0$-FePt/Fe ECC nanodots at ac-demagnetized state would be presented.
Figure 9 shows the magnetic domain structure at the remanent state after ac demagnetization of the FePt/Fe ECC bilayer samples before and after patterning. In the continuous film, magnetic multidomains were found in all the FePt/Fe ECC bilayers with various Fe thicknesses. However, the shape of the magnetic domains was irregular in the FePt single layer. With the increasing of Fe thickness, the magnetic domains became elongated and formed the random maze-like domains, which was usually found in the high-anisotropy ferromagnetic films [40]. The shape of the magnetic domains intensively depended on the film thickness especially in the continuous film with thickness less than 100 nm. The similar change in domain configuration has been reported by Okamato et al. in the FePt single layer where the magnetic domain structure varied with respect to the film thickness [35]. The strong exchange coupling between the FePt layer and Fe layer kept the FePt/Fe bilayer to have the magnetic domain structure similar to that of the magnetic single layer.

To study the magnetic domain structures of the patterned nanodots, dots with various diameters (50 nm, 100 nm, 250 nm and 500 nm) were fabricated. Here, the magnetic domain structures of the 50 nm FePt/Fe ECC dots were not shown because all were magnetic single domains. In the FePt single layer, magnetic single domain was formed in the 100 nm dots as shown in figure 9(a). As the dot size increased to 250 nm, two or more magnetic domains were formed. With the dot size further increased to 500 nm, multidomains structure was observed. This implied that the critical size of the magnetic single domain in FePt magnetically hard layer was between 100 nm and 250 nm. Theoretically, the critical size of the magnetic single domain was calculated based on a simple approximation that the domain wall energy was equal to the magnetostatic demagnetization energy. It was assumed that two magnetic domains were separated by a domain wall in the center of the rod as shown in figure 10.
Figure 9. MFM images of the FePt/Fe ECC bilayer/nanodots after ac demagnetization. The dot sizes are 500 nm, 250 nm and 100 nm in diameter. The Fe thickness varies from 0 nm (a), 2 nm (b), 3 nm (c), 4 nm (d), 6 nm (e) to 8 nm (f).

Figure 10. Schematics of a domain wall in the center of the rod which divides the rod into two equal parts with opposite magnetization directions in the out-of-plane direction.

The domain wall energy can be expressed by the following equation:

$$E_{DW} = (2Rt)\sqrt{AK_u}$$  \hspace{1cm} (2)
The demagnetization factor $N_z$ is calculated based on the following equation [36]:

$$N_z = \frac{1}{t} \left( t + R - \sqrt{t^2 + R^2} \right)$$

(3)

The demagnetization energy is:

$$E_{MS} = \frac{1}{2} NM_z^2 \left( \pi R^2 t \right)$$

(4)

where $R$ is the radius of the dot, $t$ is the height of the dot - 16 nm, $A$ is the exchange stiffness - $1 \times 10^6$ erg/cm, $K_u$ is the uniaxial magnetic anisotropy - $2.08 \times 10^7$ erg/cc, $M_s$ is the saturation magnetization - 1000 emu/cc.

Using numerical analysis, the critical single domain size of the 16 nm $L1_0$-FePt nanodots was calculated to be around 242 nm when $N$ in Equation (4) equal to $N_z$ in Equation (3). Theoretically, a small amount of demagnetization energy would remain in the dot after forming two magnetic domains. Therefore, $N$ is smaller than $N_z$, the critical single domain size would be slightly smaller than 242 nm, but still much larger than 100 nm. In addition, the edge damage caused by ion milling and intrinsic magnetic defect would promote the domain configuration transition from single domain to multidomains at a smaller critical size [41]. This explained why more than two domains were founded in some 250 nm dots.

The magnetic domain structure of the 250 nm and 500 nm dots was modified by the exchange-coupling of Fe layer. With 2 nm Fe layer, the magnetic multidomains structure remained in both 250 nm and 500 nm dots. When the Fe thickness went beyond 4 nm, magnetic multidomains in the 250 nm and 500 nm dots transferred into magnetic single domain. To carefully investigate the transition from multidomains to single domain, a bilayer sample with 3 nm Fe was fabricated. The magnetic domain structure of the FePt/Fe (3nm) nanodots was shown in figure 9(c). A transition in the magnetic domain structure with increasing the Fe layer thickness under the same dot size was clearly observed. In the 250 nm dots, the magnetic multidomains
structure was replaced by the magnetic single domain structure; while in the 500 nm dots, less dots showed the magnetic multidomains structure.

The modification of the magnetic domain structure by the Fe layer can be understood as follows. Due to the strong exchange-coupling between the Fe soft layer and the FePt hard layer, the magnetization of the top FePt layer would tilt away from the out-of-plane direction, hence reduced the demagnetization field and the total demagnetization energy. The purpose of forming magnetic multidomains is to reduce the demagnetization energy. Films having smaller demagnetization energy will have larger single domain size. As the Fe layer thickness increases, the demagnetization energy decreases further, leading to larger magnetic single domain. Furthermore, the thicker Fe soft layer would enhance the magnetic interaction in the FePt/Fe ECC nanodot. The FePt layer is not a perfect single crystal but polycrystalline in our work. The exchange-coupling strength among the adjacent single crystalline FePt grains would be less than the bulk value. Because the exchange stiffness of Fe is almost double that of the FePt layer, due to the strong exchange-coupling between the Fe layer and FePt layer, the FePt would behave like having a larger effective exchange stiffness when the Fe layer becomes thicker. This may increase the domain wall energy and forming magnetic multidomain is less preferred. The increase of the magnetic domain size by increasing the soft layer thickness had been reported experimentally by Wang [42] and theoretically by Suess [43].

4. Conclusions

$L1_0$-FePt nanodots and $L1_0$-FePt/Fe ECC nanodots with various diameters were fabricated by EBL patterning and followed by ion milling. Highly sensitive Hall effect measurement was applied to characterize the magnetic properties of the nanodots. The coercivity of the magnetically isolated nanodots increased 4-5 time due to the breaking of the lateral exchange coupling among the dots. However, ion milling induced damage to the dot edge which
deteriorated the magnetic properties at the edge region. The small amount of soft phase in the FePt nanodots revealed in the recoil loops was due to edge damage. The edge damage destroyed the coercivity reduction trend when thicker Fe soft phase was exchange-coupled to the FePt hard phase. Both angular dependent coercivity and simulations showed that the FePt/Fe ECC nanodots processed domain wall propagation type of magnetization reversal mechanism. The AHV/PHV plot revealed that the magnetization switched gradually from the out-of-plane direction towards the in-plane direction then to the reversed out-of-plane direction. Exchange-coupling the Fe soft phase to the FePt hard phase increased the critical size of magnetic single domain due to the reduction of magnetic flux coming out from the sample and the reduction of demagnetization energy.

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Figure 1.
Figure 2.
Figure 3.
Figure 4.
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Figure 10.