Magnetization reversal mechanism in perpendicular exchange-coupled Fe/L1₀–FePt bilayers

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New Journal of Physics 14 (2012) 073008 (14pp)
Received 14 February 2012
Published 3 July 2012
Online at http://www.njp.org/
doi:10.1088/1367-2630/14/7/073008

Abstract. The magnetization reversal mechanism in perpendicular soft/hard Fe/FePt exchange-coupled bilayers has been investigated as a function of the soft layer thickness ($t_{Fe} = 2, 3.5, 5$ nm) combining magnetization loops at variable angle, magnetic domain analysis by magnetic force microscopy and numerical micromagnetic simulations. The analytical model proposed in the literature can properly account for some features of the reversal mechanism, such as positive nucleation fields and the reduction of the perpendicular coercive field and remanence by increasing the soft layer thickness, but cannot satisfactorily describe the magnetization process of real systems. We showed that for a thickness of the soft layer exceeding the FePt exchange length ($\sim 2$ nm), numerical micromagnetic calculations are needed to reproduce experimental observations. Indeed, just above the coercive field, the magnetization reversal does not proceed in single step switching, as predicted by the analytical model, but according to a more complex process: evolution of nucleated magnetic domains whose magnetization is approximately along the surface normal in the hard layer and slightly out of the film plane in the soft layer, followed by rotation of Fe moments along the field direction.

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1. Introduction

The concept of composite magnetic material where hard and soft magnetic phases are exchange coupled at the nanometer scale was introduced many years ago as a potential route to improve the energy product of bulk permanent magnets by combining the high anisotropy of the hard phase and the high saturation magnetization of the soft phase [1]. Afterwards, this concept has been extended from bulk to multilayer systems which have been receiving a great deal of attention for their potential applications in different technological fields such as permanent magnets [2], information storage media (exchange-coupled media [3]) and magnetic microactuators and systems [4]. They also constitute a model system suitable to study fundamental properties of nanocomposite magnetic systems, the properties of individual layers being relatively easier to control.

Combinations of different hard and soft materials have been proposed in the literature and the influence on the magnetic behavior of various intrinsic and extrinsic properties (e.g. interface structure, thickness of the soft and hard layers, and layer microstructure) has been extensively studied and debated [5–11]. The fundamental role played by the soft layer thickness and the film morphology was experimentally and theoretically proved by some of the authors of this paper in Fe(soft)/L10–FePt(hard) perpendicular exchange-coupled systems [10–12]. The analytical model of Asti et al [12] describes the magnetization reversal of ideal and infinite Fe/FePt bilayers as a function of thickness and intrinsic physical properties of the single hard and soft phases. Two regimes were predicted, denoted as rigid magnet and exchange-spring magnet, depending on the soft layer thickness. In the rigid-magnet regime, the two layers reverse coherently, while in the exchange-spring regime the reversal of the soft phase starts first at a critical field ($H_n$), which can assume both negative and positive values, and supports the hard phase reversal that occurs at a coercive field ($H_c$), which progressively decreases with increasing soft layer thickness. In this model, the only mechanism allowed for magnetization reversal is the rotation of magnetic moments, which does not provide a full description of the reversal process in a real system, where the formation and evolution of magnetic domains can take place, being also affected by microstructural features (interface roughness, grain boundaries, voids, etc).

The aim of this work is to carefully explore the actual reversal mechanism in soft/hard Fe/FePt exchange-coupled bilayers as a function of the soft layer thickness combining magnetic domain analysis with magnetic force microscopy (MFM), angular magnetic measurements and numerical simulations of magnetization loops. Systems with suitable thickness values were prepared by radio frequency (RF) sputtering according to previous experimental results and theoretical predictions [5, 10–12].
2. Experimental

Fe(soft)/L1₀–FePt(hard) bilayers were deposited by RF sputtering. The background pressure of the sputtering system was $3 \times 10^{-8}$ mbar and the working pressure was $1.5 \times 10^{-2}$ mbar under Ar gas (99.99% purity). The FePt layers with nearly equiatomic composition and 10 nm nominal thickness were deposited at 415 °C by alternating Fe and Pt thin layers ($\sim 0.2$ nm each) onto a single-crystal MgO (100) substrate; after deposition, the films were annealed in vacuum at the same temperature for 100 min in order to favor the transition of the residual magnetically soft A1 phase to the magnetically hard L1₀ phase [13, 14]. On top of the FePt layer, Fe films with a nominal thickness of the soft layer ($t_{Fe}$) of 2, 3.5 and 5 nm (hereafter referred to as S2, S3.5 and S5) were deposited at room temperature and covered by a thin Pt capping layer in order to prevent Fe oxidation. In addition, a single L1₀–FePt film (hereafter referred to as S0) was also prepared as a reference system.

High-resolution x-ray diffraction and transmission electron microscopy (TEM) analysis have shown that using these conditions FePt films grow in the L1₀ phase and are epitaxially oriented on MgO with the orientation relationship FePt(001) $\langle 100 \rangle \parallel$ MgO(100)$\langle 100 \rangle$. The Fe layer grows epitaxially on top of FePt film, with a body-centered cubic structure and orientation relationship [110]$\parallel$[001]FePt [5, 11].

The surface morphology and magnetic domains structure were investigated by means of Dimension 3100 Atomic Force Microscope (AFM) equipped with Nanoscope IVa controller (Bruker Instrument). The surface topography was obtained in tapping mode, whereas magnetic domain images were acquired in tapping-lift mode with a lift scan height of 20 nm measured above topography. Magnetic images were acquired both in the demagnetized and remanent states.

Angular magnetic properties were investigated by using a commercial vector VSM magnetometer (vVSM, ADE-Technologies Model 10) equipped with two sets of pick-up coils perpendicular to each other and an electromagnet that can supply a maximum field of 2 T. Both sample and magnetization pick-up coils are stationary, while the field rotation is accomplished by rotating the electromagnet around the sample. Signals from the two sets of pick-up coils are combined and converted to give the components of the magnetization vector along four different directions, i.e. parallel ($y$) and perpendicular ($z$) to the sample (stationary reference system) and parallel and perpendicular to the external magnetic field $\mu_0 \vec{H}$ (rotating reference system) (figure 1). The $z$ and $y$ indexes will be used to denote the components of magnetization along the static reference frame when the external field is applied at $0^\circ < \phi < 90^\circ$, where $\phi$ is the angle between $\mu_0 \vec{H}$ and the $z$-axis.

Finally, micromagnetic simulations of room-temperature magnetization loops were performed using the commercial LLG Micromagnetics Simulator™, which provides a numerical solution (finite difference method) of the standard Landau–Lifshitz–Gilbert-Langevin (LLG-L) equation for the time evolution of magnetization in the presence of random thermal fluctuations (http://llgmicro.home.mindspring.com) [15–17].

3. Results and discussion

Figures 2(a)–(d) show room-temperature perpendicular ($\phi = 0^\circ$) and in-plane ($\phi = 90^\circ$) magnetization curves of Fe/L1₀–FePt films ($t_{Fe} = 0, 2, 3.5$ and $5$ nm) recorded along the external field direction. The main parameters deduced from their analysis are shown in table 1.
Figure 1. Schematic drawing of the vVSM measurement geometry. During the measurements the sample remains stationary, while the external field $\mu_0 \vec{H}$ rotates in the $(y,z)$-plane; $\phi$ is the angle between $\mu_0 \vec{H}$ and the film surface normal $(z)$.

Figure 2. Normalized ($M/M_s$), perpendicular ($\phi = 0^\circ$, $\bullet \cdots$) and in-plane ($\phi = 90^\circ$, $\circ \cdots$) room-temperature magnetization curves measured along the field direction for samples S0 (a), S2 (b), S3.5 (c) and S5 (d).

The single FePt layer (S0) shows a square perpendicular hysteresis loop with high reduced remanence $M_{\text{rem}}(0^\circ)/M_s = 0.99$ and strong effective anisotropy constant $K_{\text{eff}} = H_k M_s / 2 \sim 2 \text{ MJ m}^{-3}$, where $H_k$ is the anisotropy field estimated as the field where the in-plane
Table 1. Main magnetic parameters obtained from the analysis of room-temperature perpendicular and in-plane hysteresis loops.

| ID   | Fe sample (nm) | $M_s$ (MA m$^{-1}$) | $M_{rem}(0^\circ)/M_s$ | $M_{rem}(90^\circ)/M_s$ | $\mu_0 H_{n,0}$ a (T) | $\mu_0 H_{c,0}$ a (T) |
|------|----------------|----------------------|------------------------|------------------------|-----------------------|-----------------------|
| S0   | 0              | 0.69                 | 0.99                   | 0.03                   | −0.27                 | −0.27                 |
| S2   | 2              | 0.80                 | 0.88                   | 0.12                   | 0.56                  | −0.14                 |
| S3.5 | 3.5            | 1.05                 | 0.77                   | 0.22                   | 1.1                   | −0.11                 |
| S5   | 5              | 1.07                 | 0.55                   | 0.46                   | 1.3                   | −0.08                 |

a $\mu_0 H_{n,0}$ marks the point where the soft phase starts to deviate from the saturated state in the perpendicular hysteresis loop [12].

magnetization (hard-axis direction) and perpendicular magnetization curves merge. The small hysteretic contribution in the in-plane loop is probably due to the presence of a residual fraction of the cubic phase or to a tilting of a small fraction of tetragonal easy axis from the ideal perpendicular direction. The relatively low value of the perpendicular coercive field ($\mu_0 H_{c,0} \equiv \mu_0 H_{n,0} = 0.27$ T) is attributed to the morphology of this film, which consists, as shown in the AFM image (inset in figure 3(a)), of interconnected grains forming a quasi-continuous TEM analysis of samples discussed in [13], which were prepared in the same conditions, as well as by the MFM image of the sample in the virgin state (figure 3(a)), which shows the presence of perpendicular magnetic domains with an irregular pattern, extending over several grains.

A further proof of the quasi-continuous nature of the FePt layer is given by the analysis of the reversal mechanism, which was investigated by measuring the angular dependence of the switching field (i.e. $H_{sw}$ versus $\phi$) and comparing the results with the theoretical curve calculated for the Kondorsky model (figures 4(a) and (b)), which describes the magnetization reversal of continuous or highly exchange-coupled granular films [18]. The angular dependence of the switching field was determined by measuring, at different $\phi$ values, a series of easy-axis direct current demagnetization (DCD) curves using a non-conventional procedure that allows a more intuitive correspondence between experiments and simple models, as well described in [19]. The easy-axis DCD curves were collected by measuring at each $\phi$ angle the component of the remanent magnetization along the easy-axis direction (i.e. the $z$-axis) $M_{rem}^{z}$ and were used to determine the remanence coercivity $H_{r}$ (defined as the field where the remanence is equal to zero) which, according to Coffey et al [19], well represents the switching field $H_{sw}$. The easy-axis DCD curves and the angular dependence of $H_{sw}$ are reported in figures 4(a) and (b). Data have not been corrected for self-demagnetizing fields being, in our case, the quality factor [20] $Q = K_d/K_a \approx 8 > 1$, where $K_d = \mu_0 M_s^2/2$ is the stray field energy constant and $K_a$ is the second-order anisotropy constant. The comparison between experimental and theoretical curves reveals that the magnetic behavior of the hard FePt layer is well described by the Kondorsky model, confirming the quasi-continuous morphology of the film.

After depositing the Fe layer on top of the FePt film, the surface topography does not present significant differences with respect to the reference layer and the morphology does not sensibly change by increasing the thickness of the Fe layer (shown in the inset of
Figure 3. MFM images (bar: 0.200 \(\mu\)m) of the virgin state of samples S0 (a), S2 (b), S3.5 (c) and S5 (d). AFM images (bar: 0.200 \(\mu\)m) of the surface topography are reported in the inset.

Figure 4. (a) Room-temperature-normalized easy-axis DCD curves \((M_{\text{rem}}^{\phi}/M_{\text{rem}}^{0\degree})\) recorded at different \(\phi\) angles for sample S0. (b) Comparison between the Kondorsky model (—) and experimental angular dependence of normalized switching field \((H_{\text{sw}}(\phi)/H_{\text{sw}}(0\degree))\) for sample S0 (—•—).

figures 3(b)–(d)), the main effect consisting of a moderate increase in the degree of grain coalescence with increasing Fe layer thickness. In figures 3(a)–(d), the domain structure in the virgin state is also compared, showing that all samples present quite similar domain structure.
However, the sharp magnetic contrast clearly visible in the single FePt layer sensibly reduces with increasing Fe thickness. This effect was attributed to a reduction in the perpendicular stray field due to the contribution of the Fe moments, which tends to align toward the film plane for $t_{Fe} \geq 3.5$ nm as reported in [21, 22], and as discussed below.

Magnetization curves of Fe/FePt bilayers recorded along the external field direction (figures 2(b)–(d)) significantly change with increasing the Fe layer thickness, revealing the fundamental role played by the soft layer in determining the magnetic behavior of the whole system. For all the bilayers, the magnetization reversal under a perpendicular field starts at a positive field $H_{n,0}^{\text{Fe}}$ whose value increases with increasing $t_{Fe}$. Moreover, increasing the soft layer thickness also produces a gradual reduction of the perpendicular coercive field $H_{c,0}^{\text{Fe}}$ and reduced remanence $M_{\text{rem}}(0^\circ)/M_s$, the latter in conjunction with a steady increase of the in-plane reduced remanence $M_{\text{rem}}(90^\circ)/M_s$ (table 1).

Although some aspects, such as the presence of positive nucleation fields $H_{n,0}^{\text{Fe}}$ and the decrease of coercivity $H_{c,0}^{\text{Fe}}$ with increasing $t_{Fe}$, together with the increase of $H_{n,0}^{\text{FePt}}$, are in agreement with the predictions of the analytical model [12], there are still some discrepancies between experimental and calculated perpendicular magnetization curves, especially evident in samples S3.5 and S5, where the reversal process for the field just above the coercivity does not proceed as a single step switching, as predicted by the model, but with a more complex mechanism, reflecting the real nature of the system (i.e. formation and evolution of magnetic domains, and their dependence on the microstructural features).

To properly explain the reversal mechanism of Fe/FePt bilayers, angular remanence curves, transversal magnetization curves under a perpendicular field (i.e. $M_y$ versus $H$ curves), MFM images at remanence and numerical micromagnetic simulations of hysteresis loops were compared.

Angular remanence curves were recorded saturating the samples in a field of 1.9 T and then measuring the remanent moment along the field direction at different $\phi$ angles ($m_{\text{rem}}(\phi)$). In figures 5(a)–(d), the measurements for the four samples are reported. In the case of the single FePt layer, the polar figure shows the twofold symmetry typical of a uniaxial system and the maximum value of the remanence was observed at $\phi = 0^\circ$, confirming the perpendicular orientation of magnetization in the L1$_0$–FePt layer. Such a symmetry is still retained at the lowest Fe layer thickness (sample S2), suggesting a good out-of-plane alignment of hard and soft moments at remanence. Deviations from the twofold symmetry are observed for sample S3.5 and, for larger soft layer thickness (sample S5), the symmetry becomes fourfold and the maxima of remanence shift up according to $\phi \approx \pi/4 + n\pi/2$ ($n = 0, 1, \ldots$). The observed change in the symmetry can be explained considering that, in samples S3.5 and S5, the soft layer thickness exceeds the FePt exchange length $\lambda_{\text{ex,FePt}} = \sqrt{A/K_a} \sim 2$ nm, where $A = 1 \times 10^{-11}$ J m$^{-1}$ is the FePt exchange stiffness constant [23]. When $t_{Fe} \leq \lambda_{\text{ex,FePt}}$, due to the strong coupling effect, the magnetization of the Fe layer would be aligned along the direction normal to the film plane. At Fe thickness exceeding FePt exchange length, the magnetostatic effects become predominant forcing the Fe moments toward the in-plane direction. Indeed, in a simplified model, the total remanent moment along the field direction in a bilayer system is the vector sum of the remanent moment of each layer [24], i.e.

$$m_{\text{rem}}(\phi) = m_{\text{rem}}^{\text{Fe}} \cos(\phi-\phi_{\text{Fe}}) + m_{\text{rem}}^{\text{FePt}} \cos(\phi-\phi_{\text{FePt}}),$$

where $\phi_{\text{Fe}}$ ($\phi_{\text{FePt}}$) is the angle between the easy axis of the Fe (FePt) layer and the normal to the film surface (i.e. the $z$-axis) and $m_{\text{rem}}^{\text{Fe}}$ ($m_{\text{rem}}^{\text{FePt}}$) is the remanent moment of the Fe (FePt) layer.
Figure 5. Room-temperature angular remanence curves (experimental: –○–, fit: ——) for samples S0 (a), S2 (b), S3.5 (c) and S5 (d).

Data fit based on equation (1) allows determining the values $\phi_{\text{Fe}}$ and $\phi_{\text{FePt}}$ and then extracting information on the direction of the easy axis in each layer. For sample S2 ($t_{\text{Fe}} < \lambda_{\text{ex}}$), the fit gives $\phi_{\text{Fe}} = 15^\circ$ and $\phi_{\text{FePt}} = 0^\circ$, indicating a small tilt of the Fe moments at remanence. For samples S3.5 and S5 ($t_{\text{Fe}} > \lambda_{\text{ex}}$), $\phi_{\text{Fe}} = 90^\circ$ and $\phi_{\text{FePt}} = 0^\circ$ (i.e. at remanence, FePt and Fe moments preferentially lie along directions perpendicular and parallel to the film plane, respectively) and the different positions of maxima in the two angular remanence curves are related to the different $m_{\text{rem}}^{\text{Fe}}/m_{\text{rem}}^{\text{FePt}}$ ratios for the two samples.

The rotation of the Fe moment toward the film plane is proved further by the presence of an in-plane magnetization component ($M_y$) in hysteresis cycle measurements under a perpendicular field in the bilayers (figures 6(b)–(d)), whereas no $M_y$ signal was detected in the whole $H$-range in the single FePt layer (figure 6(a)). As a soft Fe layer is deposited on top, the $M_y/M_s$ signal increases on moving from the saturated state to the remanent state, confirming that the Fe moments rotate toward the in-plane direction. The increase in $M_y/M_s$ is indeed very small for $t_{\text{Fe}} = 2$ nm and becomes larger when $t_{\text{Fe}} > \lambda_{\text{ex}}$, in agreement with our previous comments.
The competition between the interlayer exchange coupling and the magnetostatic effects is expected to strongly affect the magnetization reversal mechanism. For a soft layer with a thickness of 2 nm, in a hysteresis cycle measurement under a perpendicular field, the Fe moments remain essentially aligned along the easy axis of the L1₀-FePt layer and the bilayer exhibits rigid-magnet-like behavior, although MFM investigation (figure 7) indicates that the magnetization reversal occurs through nucleation and the evolution of perpendicular reverse domains (within the whole system) and not by moment rotation as proposed by the analytical model [12].

By increasing the soft layer thickness, the enhanced magnetostatic contribution combined with the interlayer exchange coupling produces an increase in the positive nucleation field as well as a decrease in the perpendicular reduced remanence and coercive field, and the bilayer, as a whole, behaves like an exchange-spring system. However, the overall magnetization reversal process cannot be satisfactorily described by the model, thus, a numerical micromagnetic calculation of the perpendicular hysteresis loop of sample S5 ($t_{Fe} = 5$ nm) was performed to better describe the actual reversal process (figures 8 and 9).

Figure 8 shows experimental and simulated data for the components of magnetization parallel to the film surface ($M_y/M_s$) and to the external field ($M/M_s$) applied along the z-axis. The LLG-L equation was solved by subdividing the simulated volume ($650 \times 650 \times 16$ nm$^3$).
Figure 7. MFM image (bar: 4.0 µm) of the remanent state for sample S2; data have been obtained after saturating the sample in a field perpendicular to the film plane.

Figure 8. Comparison between experimental data and numerical micromagnetic simulation of magnetization parallel ($M/M_s$) and perpendicular ($M_y/M_s$) to the applied field for the perpendicular hysteresis cycle of sample S5. (– – – –): Experimental data; (—–): simulated curve with interlayer exchange coupling $A_{\text{int}} = 0.4 \times 10^{-11}$ J m$^{-1}$, and anisotropy constant $K_a = 1.0$ MJ m$^{-3}$; (– – – –): simulated curve with $A_{\text{int}} = 0$. The micromagnetic states of selected points (labeled $a$–$f$) on the hysteresis loops are shown in figure 9.
Figure 9. Evolution of simulated micromagnetic states during the application of the magnetic field in terms of the cosine direction of local magnetization (i.e. the components of the magnetization parallel to the external field $M/M_s$ and to the film surface $M_y/M_s$) of both soft (Fe) and hard layers (FePt); color scale: red $M/M_s = 1$, black $M/M_s = 0$, blue $M/M_s = -1$. The MFM image of the remanent state (figure 10) shows the presence of perpendicular reverse domains and bigger elongated stripe-like domains with a reduced contrast, possibly due to a slight tilting of the magnetization vector away from the surface plane.

with a regular mesh of prismatic cells ($256 \times 256 \times 8$ in our case, with a volume $V = 2.5 \times 2.5 \times 2.0$ nm$^3$ for each cell). The simulated thickness value (i.e. a multiple of the FePt exchange length $\lambda_{ex,FePt} = \sqrt{A/K_a} \sim 2$ nm, yielding an integer number of sublayers) was chosen to provide a computationally efficient discretization scheme without sacrificing calculation accuracy, given the computation facility available.

To allow a meaningful comparison with experimental results, material parameters such as saturation magnetization $M_s$ and uniaxial magnetic anisotropy $K_a$ were chosen consistently with the experimental values determined by VSM measurements. Namely, for FePt, $M_s = 0.75$ MA m$^{-1}$ and $K_a = (0.5–1.0)$ MJ m$^{-3}$; a value of $1 \times 10^{-11}$ J m$^{-1}$ was chosen for the exchange stiffness constant $A$ smaller than $2 \times 10^{-11}$ J m$^{-1}$ which is expected for Fe [23]. The interlayer Fe/FePt exchange coupling ($A_{int}$) was left as a free parameter in order to give the best
Figure 10. MFM image (bar: 4.0 µm) of remanent state for sample S5; data have been obtained after saturating the sample in a field perpendicular to the film plane.

accordance between simulation and experiment, i.e. $A_{\text{int}} = 0.4 \times 10^{-11} \text{ J m}^{-1}$, in our case, which is a plausible value considering the interface roughness and waviness [5, 10].

The simulation well reproduces the field dependence of both magnetization components, with a satisfactory agreement between calculated and experimental reduced remanence and coercive field values. The calculation, however, tends to underestimate the field where the loop closes in the third quadrant. Such a deviation may be attributed to the actual material microstructural features. Figure 8 also shows the hysteresis loop for a fully exchange decoupled Fe/FePt bilayer ($A_{\text{int}} = 0$), which does not reproduce the experimental curve confirming the important role played by the interface coupling.

The LLG-L simulation software allows tracking of the micromagnetic state during the application of the magnetic field in terms of the direction cosines of local magnetization (i.e. the components of magnetization parallel to the film surface $-M_y/M_s$ and to the external field $-M/M_s$) of both soft and hard layers (figure 9), thus providing detailed information on the magnetization reversal mechanism. It has been verified that the simulated magnetization distribution does not depend on the $z$-coordinate (i.e. it does not depend on the specific sublayer). Hence, figure 9 shows, as an example, the evolution of simulated micromagnetic states during the application of the magnetic field for an FePt sublayer in the middle of the simulated volume and the uppermost Fe sublayer.

Initially, the sample is fully saturated in the $z$-direction beyond $\mu_0 H_{\text{sat}} = 2$ T (full red scale). As a reverse field is applied, in the range $0 < H < H_a$ (point $a$), an in-plane component of magnetization appears in the soft layer, confirming the rotation of the Fe moments toward the in-plane direction due to the magnetostatic effects. In the same $H$-range, the FePt moments are not affected by the reverse field and remain essentially oriented along the direction of the saturation field. At the sharp edge in the magnetization curves (point $b$), reverse regions form in the soft layer with the magnetization slightly tilted in the out-of-plane direction; correspondingly, in the hard layer, reverse regions form with the magnetization highly oriented in the perpendicular
direction, thus indicating that the Fe phase supports the nucleation processes of the hard phase leading to a reduction in the coercive field (point c) as expected for an exchange-spring system.

The micromagnetic simulation is able to well reproduce the presence of the perpendicular reverse domains, whereas the large elongated domains are too wide to be handled by the computation facility available and should be related to the Fe moment lying very close to a direction parallel to the film plane.

Above the coercive field (point d–e) there is a change of the magnetization curve slope, which was not predicted analytically and could be related to a complex evolution of magnetic domains whose magnetization is approximately along the surface normal in the hard layer and slightly out of the film plane in the soft layer. The reversal process ends up with the saturation of the hard phase followed by the rotation of the Fe moments along the field direction (point f).

4. Summary and conclusions

The magnetization reversal mechanism in soft/hard Fe/FePt exchange-coupled bilayers with different soft layer thickness has been investigated combining magnetization loops at variable angle, MFM magnetic domain analysis and numerical micromagnetic simulations. It turns out that the analytical model proposed by Asti et al can properly account for some features of the reversal mechanism, such as positive nucleation fields and the reduction of the perpendicular coercive field and reduced remanence by increasing the soft layer thickness. Nevertheless, in our system, where magnetic domains are present and also affected by microstructural features, for thickness of soft layer larger than the FePt exchange length, a more complete description of experimental magnetization curves can be given by numerical micromagnetic calculations. It has been shown that, just above the coercive field, the magnetization reversal does not proceed in single step switching as predicted by the model, but in a more complex process: evolution of nucleated magnetic domains whose magnetization is approximately along the surface normal in the hard layer and slightly out of the film plane in the soft layer, followed by rotation of Fe moments along the field direction.

Acknowledgments

The authors thank E Patrizi for technical assistance in magnetic measurements. This work was supported by the European Commission FP7 project TERAMAGSTOR (contract no. FP7-ICT-2007-2-224001) and by the MIUR project PRIN08 (Thermal stability of exchange-spring planar magnetic nanostructures with perpendicular and lateral exchange coupling).

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New Journal of Physics 14 (2012) 073008 (http://www.njp.org/)