Magnetic anisotropy axis reorientation at ultrathin FePt films

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Ultrathin FePt films (thickness between 1 nm and 5 nm) were studied for non-volatile memories applications. The films were magnetron sputtered on monocrystalline MgO $\langle 001 \rangle$ substrates at 500 °C. The films are polycrystalline, except the 1 nm thick film which is not continuous. It is shown that films with thickness higher than 2.7 nm have L1₀ structure and perpendicular magnetic anisotropy, while a transition to in-plane anisotropy occurs for thinner films. The out-of-plane coercivity drops from 16 kOe at the thicker film to 0.5 kOe at the thinner one.





1 Introduction FePt thin films are studied since several years for applications in magnetic recording [1, 2] and spintronics [3-5]. A recently proposed magnetic memory design makes use of spin-orbit torques for reversing the magnetization of ultrathin ferromagnetic films [6] and FePt has been suggested as a candidate ferromagnetic material [7]. However, ferromagnetic layers with thickness between 1 nm and 2 nm, or even in the sub-nm range, are required in this case [6]. What is more, perpendicular magnetic anisotropy is necessary for obtaining a high areal density memory, as more bits can be packed at the same area. In this work, sputtered FePt films with thickness between 1 nm and 5 nm are investigated as candidate materials for non-volatile magnetic memories applications and their structural, morphological, and magnetic properties are studied.

2 Experimental details Films have been deposited on monocrystalline MgO(001) substrates using an ultrahigh vacuum magnetron sputtering setup. The substrates were kept at 500 °C during deposition for promoting the conversion from the magnetically soft fcc phase to the magnetically hard fct phase with alternate stacking of (001) planes of Fe and Pt (usually called $L1_0$ phase) [8]. Higher temperatures were avoided for maintaining compatibility to microelectronics processes. The source material was a Fe₅₀Pt₅₀ sputtering target. All films were capped with a 3 nm thick Ta layer for protection against oxidation. X-ray diffraction (XRD) structural analysis has been performed using a Siemens D500 diffractometer with Cu K_a radiation. Anomalous Hall Effect (AHE) measurements at room temperature have been conducted with the applied magnetic field perpendicular to the samples surface. Atomic Force Microscopy (AFM) and Magnetic Force Microscopy (MFM) have been performed using a Bruker Dimension Icon microscope and commercial probes.

3 Results and discussion

3.1 Structure and microstructure Figure 1 shows θ - 2θ scans obtained at the studied samples. The diagrams





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Figure 1 XRD θ -2 θ scans of the studied samples. Diagrams have been sifted at the vertical scale for clarity purposes.

are dominated by the strong Bragg peaks produced by the MgO(001) substrate. The fundamental (002) and (004) and the superlattice (001) and (003) peaks of the L1₀ FePt structure are clearly seen at the XRD diagrams and are marked in the figure by vertical solid lines. The peak at around 48.6° is a convolution of the Bragg peaks of the (200) fcc (at 47.28°) and (002) L1₀ (at 49.18°) phases and it is shifted towards the (200) peak as film thickness becomes smaller. It should be noted that the Bragg reflection of the fcc FePt(110) planes at 33° is not apparent. The lower intensity peaks at around 29.2°, 35.4°, 38.6°, 41.1°, and 82.8° correspond to Ta or Ta₂O₅.

For calculating the chemical ordering parameter S of the films, the intensity of the (001) and (002) peaks is measured. S may be calculated by the following formula [8]:

$$S \cong 0.85 \left(\frac{I_{001}}{I_{002}}\right)^{1/2},\tag{1}$$

where I_{001} and I_{002} are the integrated intensities of the corresponding XRD peaks, calculated from least-squares pseudo-Voigt fitting [9]. The deconvolution of the mixed (200) and (002) peak yields the integrated area of the L1₀ (002) peak (for more details see Supporting Information). *S* is unity for perfect order and zero when the atoms have been randomized sufficiently for the layered structure to be indistinguishable over more than a few interatomic distances [8]. The obtained values are shown in Table 1.

Table 1 FePt parameters (*S*, $M_{\rm R}/M_{\rm S}$, $H_{\rm C}$) as a function of film thickness.

film thickness (nm)	S	$M_{ m R}/M_{ m S}$	H _C (kOe)
1	0.54	0.27	0.5
1.5	0.50	0.45	3
2	0.62	0.90	11.7
2.7	0.69	0.99	14.8
5	0.70	1.00	16

There is a clear decreasing trend of the S parameter as the thickness decreases, indicating lower ordering as film thickness becomes lower. However, it should be noted that S values may be poorly defined for the thinnest films, for which their total thickness is only a few interatomic distances. Thus, we consider the obtained S values as a qualitative guide of the films' chemical order.

Figure 2 shows surface morphology images obtained using AFM. Although the actual surface imaged is the one of the top Ta layer, as its thickness remains constant for all samples, variations of the obtained morphology are attributed to the underlying FePt film. The measured surface roughness remains relatively unchanged for FePt thickness between 1.5 nm and 2.7 nm, while it increases at the 5 nm thick film, as expected due to its higher thickness. However, a steep surface roughness increase is observed at the 1 nm thick sample: the obtained value is roughly two times higher than that of the 1.5 nm thick sample. This result indicates that the thinnest sample is not a continuous FePt layer and it represents the initial stages of FePt island growth on MgO. The lateral size of the observed features, as measured by AFM imaging, is also maximum at the thinnest film, in accordance to the island FePt morphology of this sample; however, it remains relatively unchanged for the rest of the samples, between 10 nm and 15 nm. The later suggests that variations of the film magnetic properties (see next section) depend on grain structure, as film microstructure does not changes considerably.

3.2 Magnetic properties Magnetic hysteresis cycles were obtained by means of AHE measurements. We employ this method because the measurement signal is inversely proportional to the film thickness, as it actually depends on the film's sheet resistivity; this makes possible to



Figure 2 Surface morphology imaging using AFM. The FePt layer thickness is (a) 1 nm, (b) 1.5 nm, (c) 5 nm. The color scale is 6.5 nm (from dark to bright) and the scale bar is 600 nm, in all images. (d) RMS surface roughness and average size of lateral features as a function of FePt layer thickness, obtained from several AFM images of each sample.

obtain hysteresis loops even at the thinnest samples, without the need of low temperatures or expensive equipments.

For the measurement geometry that we used (applied field perpendicular to the sample plane), the measured Hall voltage $V_{\rm H}$ is related to the applied field H and the perpendicular component of the sample magnetization $M_{\rm P}$ according to the following formula [11]:

$$V_{\rm H} = \frac{R_{\rm H}I}{t}H + \frac{\mu_0 R_{\rm S}I}{t}M_{\rm P} , \qquad (2)$$

where *I* is the current, *t* is the layer thickness, μ_0 is the vacuum permeability, $R_{\rm H}$ is the ordinary Hall coefficient, and $R_{\rm S}$ is the anomalous Hall coefficient. The ordinary Hall voltage is negligible compared to the anomalous Hall voltage. However, as the normal Hall effect is linearly dependent to *H*, it superimposes a small slope on the hysteresis loop, which is negative in our measurements as the net majority carriers are electrons. For removing this constant slope the straight part of the hysteresis loops after magnetization saturation is fitted and the obtained curve is subtracted by the data.

The AHE hysteresis loops are shown in Fig. 3. A clear evolution of magnetic properties is observed: both remanence magnetization (M_R) and the out-of-plane coercive field $(H_{\rm C})$ decrease as FePt film thickness decreases. It should be noted that $M_{\rm R}$ is determined from the normalized Hall voltage value at zero applied magnetic field, while the out-of-plane coercive field $H_{\rm C}$ equals the magnetic field value where the normalized Hall voltage drops to zero (no $M_{\rm P}$ component). Moreover, the loop squareness, defined as $M_{\rm R}/M_{\rm S}$, is measured (where $M_{\rm S}$ is saturation magnetization). It should be noted that only the $M_{\rm R}/M_{\rm S}$ ration may be determined quantitatively, as AHE does not measures sample magnetization. The obtained values are shown in Table 1. It can be seen that the two thicker films have clearly perpendicular magnetic anisotropy, in accordance to their high chemical ordering [11]. A perpendicular to in-plane mag-



Figure 3 AHE hysteresis loops of all the studied samples obtained at room-temperature. Legend indicates the FePt layer thickness.



Figure 4 (a) MFM phase image of the 5 nm thick FePt film. The color scale is 1.4 deg. (from dark to bright), the lift-height during acquisition was 40 nm and the scale bar is 600 nm. The lift-height was 40 nm. (b) FFT of such MFM image (inset) and its radial average in logarithmic scale.

netic easy axis transition occurs as the FePt thickness becomes lower and the contribution from the fcc FePt grains increases; for the thinnest film the perpendicular axis is the hard magnetization axis. It should be noted that the small magnetic field needed to saturate the thinnest sample at room temperature is in contradiction to the existence of any superparamagnetic phase. The $M_{\rm R}/M_{\rm S}$ and out-of-plane $H_{\rm C}$ values decrease consistently with the decreasing S value.

Figure 4(a) shows a representative MFM image of the 5 nm thick simple in the demagnetized state. Low signalto-noise ratio prohibited imaging the magnetic structure of thinner samples. Dark and bright regions forming an island-like pattern can be observed, and they can be ascribed to domains with up or down magnetization. Figure 4(b) shows the Fast Fourier Transform (FFT) of the same MFM image and its radial average in logarithmic scale. A clear peak located at 1.83 µm⁻¹ is distinguished, which corresponds to a 546 nm periodicity in the real space. Thus, the average width of each dark or bright region is half that value, i.e. 273 nm. This large value is not unexpected, as it has been shown that magnetic domain size of thin films with perpendicular magnetic anisotropy increases as thickness decreases [12, 13]. The obtained size is several times bigger than that deduced for 4 nm thick FePt granular films (approximately 20 nm) that exhibited weak magnetic coupling between grains [14], which pinpoints to a strong magnetic coupling in our case. This is in agreement with the evolution reported for thicker films, where the islandlike domains coalesce and a maze-like pattern is formed [15]. Concerning the domain walls, their accurate profile could not be deduced from our experiments due to the convolution with the MFM probe, but from micromagnetic simulations only a few nm width is expected [16].

4 Conclusions Concluding, magnetron sputtered ultrathin FePt films have been studied, with thickness between 1 nm and 5 nm; their structural, microstructural, and magnetic properties are presented and discussed. All films are polycrystalline, except the 1 nm thick film which is not continuous. Films with thickness equal or higher than



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2.7 nm have $L1_0$ structure and perpendicular magnetic anisotropy. An out-of-plane to in-plane magnetic anisotropy transition occurs as FePt thickness is decreased. This strong dependence of anisotropy and magnetic coercivity on film thickness could allow the fabrication of pseudospin-valves with a thicker "pinned" $L1_0$ FePt layer and a thinner "free" FePt layer for sensors or non-volatile magnetic memories applications.

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Supporting Information Additional supporting information may be found in the online version of this article at the publisher's website.

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