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The temperature-dependent in- and out-of-plane magnetic anisotropies in $Fe_n/V_m(001)$ superlattices

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Abstract. $Fe_n/V_m(001)$ superlattices grown on MgO(001) are studied by ferromagnetic resonance (FMR). The temperature dependence of the second- and fourth-order in- and out-of-plane magnetic anisotropies were determined from azimuthal and polar angular-dependent FMR for an Fe_4/V_4 sample. A very small step-induced in-plane uniaxial anisotropy $K_{2\parallel}$ indicates the existence of very large atomically flat terraces resulting in a perfect structural quality, in agreement with x-ray diffraction studies. The competition between surface and volume anisotropic components results in an unusual temperature dependence of the total in-plane fourfold anisotropy $K_{4\parallel}$ possibly changing the easy axis from the [100] to [110] in-plane directions at temperatures higher than room temperature. Finally, the FMR-linewidth values are typical for samples with a very high degree of homogeneity, comparable to that of Fe whiskers.

1. Introduction

Magnetic multilayers have attracted a lot of interest in recent years, mainly because of their technologically important properties such as perpendicular anisotropy [1] and giant magnetoresistance [2]. Less work has been devoted to the study of the in-plane anisotropy. This was reasonable, since the majority of the multilayered systems were polycrystalline structures with a varying degree of texture and consequently they presented no in-plane magnetic anisotropy. Furthermore, the total thickness was usually large resulting in many cases in columnar growth with a lot of structural imperfections such as twins or misfit dislocations [3]. The evolution of the preparation methods has, nowadays, made it possible to grow high-quality epitaxial multilayers. Such multilayers have been proven to fulfil the criteria for calling them superlattices [4]. Their single-crystalline nature allows the development of an in-plane magnetic anisotropy as well. The study of the in-plane anisotropy is important for a thorough understanding of the mechanism of the magnetic anisotropy itself and the revelation of possible in-plane spin reorientation phase transitions which up to now have mostly been studied in epitaxial ultrathin films due to their structural perfection [5].

The structural, magnetic and superconducting properties of textured polycrystalline {110} Fe/V multilayers were studied during the past 1980s [6]. Less attention has been

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paid to this system during this decade until recently. In this contribution, high-quality Fe_n/V_m superlattices have been prepared on MgO(001) by UHV-based sputtering and the temperature dependence of the magnetic anisotropies has been studied by FMR. (n and m denote the number of monolayers (ML) of the constituents within each modulation period.) An antiferromagnetic-like exchange coupling between the Fe layers has been clearly demonstrated via hysteresis loops and FMR for $m = 13$ ML and preliminary magnetic anisotropy studies have been reported for this series of Fe_n/V_m superlattices [7, 8]. Here, we will present a systematic FMR study for an Fe_4/V_4 superlattice. In such a sample the Fe-layer thickness is near the one for an in-plane isotropic behaviour at room temperature [7].

This paper is organized as follows. First, in section 2, a detailed description of the experimental method is given. In section 3 the full determination of all anisotropy parameters together with their temperature dependence is presented. The appearance of a very small in-plane uniaxial anisotropy reveals the presence of extremely large atomically flat terraces. The fourfold in-plane anisotropy depends dramatically on temperature and an in-plane reorientation of the easy-magnetization axis is expected to occur at about 400 K. Such a trend has also been recorded for an Fe_3/V_8 sample. The shape anisotropy is far larger than any other out-of-plane anisotropy, so that one does not expect an out-of-plane reorientation at any Fe thickness or temperature. In section 4 we analyse the FMR linewidth, its homogeneous and inhomogeneous contributions, from which we can deduce the high quality of the structure, in direct comparison with that of Fe whiskers.

2. Experimental details

The samples were produced in a three-source ultra-high-vacuum- (UHV-) based sputtering equipment and their high structural quality has been revealed by x-ray diffraction as described in detail elsewhere [4, 7]. The Fe_n/V_m superlattices have been grown pseudoepitaxially on the MgO(001) with their [100] in-plane direction coinciding with the [110] of the substrate. The Fe_4/V_4 sample has 40 periods with a modulation wavelength of 1.177 nm which results in a total thickness of 47.1 nm.

The study of the magnetic anisotropies has been carried out via FMR at the frequencies of about 4 and 9.2 GHz in the temperature range 10–300 K. For the hard-axis magnetic out-of-plane measurements large magnetic fields are needed. Our maximum available field of 22 kOe was adequate to measure the resonance field H_r in this direction. The determination of the temperature dependence of the saturation magnetization M of the sample is necessary for a separation of the anisotropy constants, as will be described later on. M was measured by vibrating sample magnetometry (VSM) along the [110] in-plane direction of Fe in an applied magnetic field of 500 Oe capable of bringing the sample to magnetic saturation.

Ferromagnetic resonance is a sensitive tool for the investigation of magnetic materials. It allows a precise determination of the anisotropy constants from angular-dependent measurements of the H_r [5, 9–11]. In figure 1 the FMR spectra, corresponding to the field derivative of the absorbed microwave power, are shown in the temperature range between 10 and 290 K at a frequency of 9.2 GHz. The magnetic field was applied along the Fe-[110] in-plane direction. As one may see, the spectra are noisefree and their shape is almost ideal Lorentzian. H_r is determined as the field found at half the distance between the two maxima of each spectrum. In figure 1 H_r shifts towards smaller values with increasing temperature, because, as we will see in section 3, the magnetic field is applied along the hard in-plane direction. In the present work we have performed two types of angular-dependent measurement: first, in plane by varying the azimuthal angle φ_H measured from the Fe-[100]

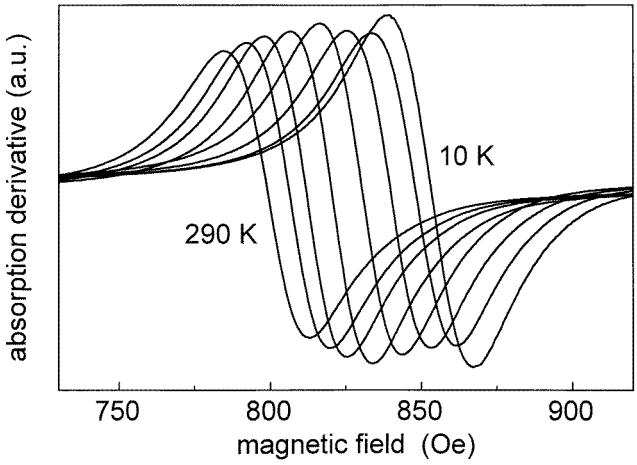

Figure 1. Spectra of the absorption derivative of the FMR microwave power for an Fe_4/V_4 superlattice on MgO(001) substrate, measured at 9.2 GHz with the static field applied along the [110] direction. The temperature from the left (290 K) to the right (10 K) decreases by constant steps of 40 K.

direction; secondly, out of plane as a function of the polar angle θ_H measured from the film normal [001] to the in-plane [110] direction of Fe.

The distance between the two maxima of each spectrum in figure 1 defines the FMR linewidth ΔH_{exp}. Frequency-dependent measurements of ΔH_{exp} allow an evaluation of the Gilbert parameter G. Angular-dependent ΔH_{exp} studies also provide information about the inhomogeneity of the magnetic material. The magnitude of the FMR linewidths is a measure of the magnetic and structural quality of the samples.

3. Anisotropies

3.1. Determination of the magnetic anisotropies

The magnetic anisotropy of a tetragonal system is usually described by a free energy density including second-order out-of-plane $K_{2\perp}$, fourth-order out-of-plane $K_{4\perp}$ and in-plane $K_{4\parallel}$ anisotropy constants [10]. Figure 2 shows the angular dependences of H_r (at 9.23 GHz and 10 K) on φ_H (figure 2(a)) and θ_H (figure 2(b)). The first dependence allows us to obtain the anisotropy constants $K_{2\perp}$ and $K_{4\parallel}$ from the resonance condition in plane [5, 10]:

$$\left(\frac{\omega}{\gamma}\right)^2 = \left[H_{r\parallel}\cos(\varphi-\varphi_H)+2\frac{K_{4\parallel}}{M}\cos 4\varphi\right]\left[H_{r\parallel}\cos(\varphi-\varphi_H)+2\left(2\pi M-\frac{K_{2\perp}}{M}\right)+\frac{K_{4\parallel}}{2M}(3+\cos 4\varphi)\right]. \quad (1)$$

In principle, (1) can be modified to provide a value for the in-plane uniaxial anisotropy $K_{2\parallel}$ [12, 13], which is very small for our film, as we will see later on. The second dependence makes feasible a calculation of the out-of plane constant $K_{4\perp}$ from the resonance condition

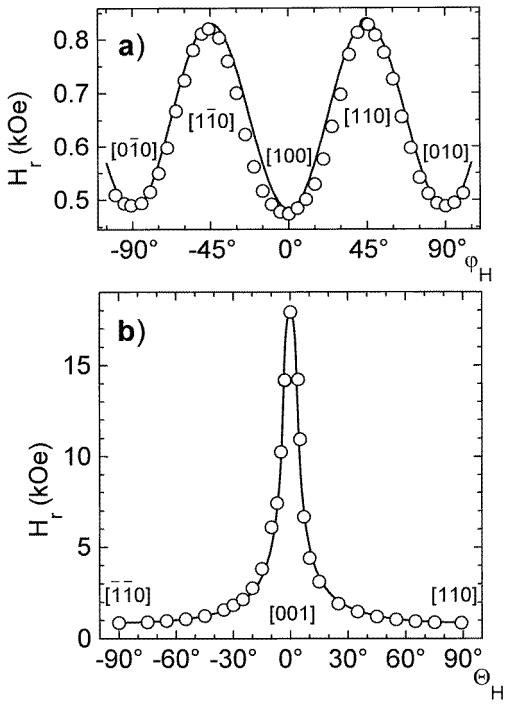

Figure 2. Angular dependences of H_r on the external magnetic field direction: (a) in plane φ_H (measured from the [100] axis) and (b) out-of-plane θ_H (from the out of plane [001] to the in-plane [110] axis). Frequency: 9.23 GHz. Temperature: 10 K. Circles denote experimental data. Solid lines are the result of a fit procedure with $2\pi M - K_{2\perp}/M = 6.75$ kOe, $K_{4\parallel}/M = 0.095$ kOe and $K_{4\perp}/M = -0.63$ kOe.

for the film planes {110}:

$$\left(\frac{\omega}{\gamma}\right)^2 = \left\{H_r\cos(\theta-\theta_H) - \left[2\left(2\pi M - \frac{K_{2\perp}}{M}\right) - \frac{K_{4\perp}}{M} + \frac{K_{4\parallel}}{2M}\right]\cos 2\theta + \left(\frac{K_{4\perp}}{M} + \frac{K_{4\parallel}}{2M}\right)\cos 4\theta\right\} \times \left\{H_r\cos(\theta-\theta_H) - \left[2\left(2\pi M - \frac{K_{2\perp}}{M}\right) + \frac{K_{4\parallel}}{M}\right]\cos^2\theta + \left(\frac{2K_{4\perp}}{M} + \frac{K_{4\parallel}}{M}\right)\cos^4\theta - \frac{2K_{4\parallel}}{M}\right\} \quad (2)$$

by using the values of $K_{2\perp}$ and $K_{4\parallel}$ determined via (1).

Generally, (2) is adequate for the evaluation of all three anisotropy constants when the γ value is known [9–11]. In such a fit procedure $K_{2\perp}$ affects mostly the $H_{r\parallel}$ value, $K_{4\perp}$ the $H_{r\perp}$ value, while $K_{4\parallel}$ has a slight effect on $H_{r\parallel}$ and a particular influence on the shape of $H_r(\theta_H)$. However, a combination of the analysis of both in- and out-of-plane angular dependences results in a precise determination of the K-values. Furthermore, an evaluation of the g-factor is also possible through this fitting procedure, when $(\omega/\gamma)^2$ plays an essential role in (1) and (2). While this condition holds well for Ni (and similar results for ultrathin Ni/Cu(001) films will be published in the near future), Fe possesses a large magnetization

and thus an evaluation of the g-factor value through (2) for Fe-based samples is questionable. Consequently, for the analysis of our results we have used $g = 2.09$, a value determined previously for Fe/MgO(001) [14] and in good agreement with Fe-containing structures found in literature [15–17].

A careful inspection of figure 2(a) reveals a slight asymmetry between the two ideally identical [100] and [010] directions. This asymmetry is clearly depicted in figure 3 ($T = 300$ K) because of the smaller range of the H_r-axis. Such an asymmetry indicates a small in-plane uniaxial anisotropy $K_{2\parallel}$. Its magnitude is almost one order of magnitude smaller than the $K_{4\parallel}$, which is already very small, as we will show later on. The appearance of this anisotropy term could possibly be related to step-induced anisotropy. Such an anisotropy has been recently studied for ultrathin bcc Fe/Ag(001) [18]. If one uses the step-induced anisotropy constant K_{sp} from [18], a very large terrace size of about 2000 nm is obtained for our sample, while the easy axis runs parallel with the step edges. Such a large terrace size is difficult to realize in practice; however, this outcome is in qualitative agreement with the ideal-like low- and high-angle x-ray diffraction data of the Fe_n/V_m superlattices [4, 7]. Let us note, finally, that FMR is one of the most sensitive techniques in revealing the existence of such small anisotropies.

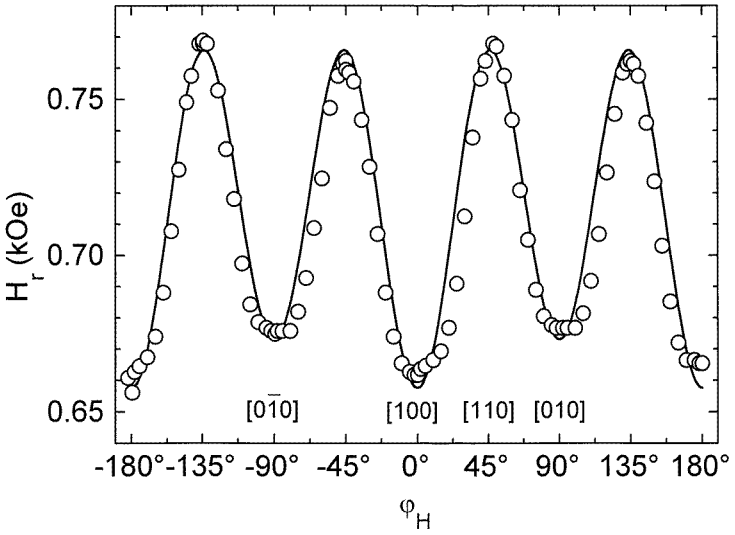

Figure 3. Angular dependence (circles) of H_r on the in-plane angle φ_H at 9.23 GHz at 300 K. The numerical simulation (solid line) is carried out with the fit parameters as follows: $2\pi M - K_{2\perp}/M = 6.58$ kOe, $K_{4\parallel}/M = 0.026$ kOe, $K_{2\parallel}/M = 0.004$ kOe.

3.2. The temperature dependence of the magnetic anisotropies

As shown in (2), by using the H_r values for the three directions, in-plane [100] and [110] and out-of-plane [001], one may obtain the parameters $2\pi M - K_{2\perp}/M$, $K_{4\parallel}/M$ and $K_{4\perp}/M$. Furthermore, from $H_r(T)$ measurements for these directions it is possible to determine the temperature dependence of these parameters. For a separation of the K-constants from M the temperature dependence of the magnetization is necessary. However, since the $M(T)$ for Fe-based alloys or multilayers is very different from the $M(T)$ of pure Fe (see, for example, [6]), one should actually measure the $M(T)$ for our sample with a static experiment.

We have measured the $M(T)$ curve via vibrating sample magnetometry (VSM) in the temperature range 3–200 K (figure 4). The FMR absorption line intensities, which result from the double integration of the Lorentzian-type spectra shown in figure 1, are an additional measure of the quantity $M(T)/M(0)$, where $M(0)$ stands for the saturation magnetization at 0 K [19, 20]. Such results (normalized to $M(T)$) are also shown in the same figure, for comparison. There is a good agreement between these two methods of measuring the magnetization. The slight reduction of M between 3 and 200 K suggests that the Curie temperature (T_C) of our sample is far above the temperature range of measurements. In such a case the experimental $M(T)$ dependence can be approximated by a Bloch law of the form $4\pi M(T)/4\pi M(0) = 1 - \beta T^{3/2}$, where $4\pi M(0) = 16.2$ kOe and $\beta = 1.45 \times 10^{-5}$. The reduction of the $4\pi M(0)$ of our sample in comparison with the bulk Fe ($4\pi M(0)_{Fe} = 21.55$ kOe) is an effect of the Fe–V interface and it is discussed in detail elsewhere [7]. The β-value, determined by a least-squares fit, is larger than the corresponding one for bulk Fe, $\beta_{Fe} = 0.932 \times 10^{-5}$ [21], or that of thick Fe films (e.g. for 6 nm Fe/W(110) $\beta_{Fe/W} = 0.974 \times 10^{-5}$ [22]) but is in good agreement with measurements on polycrystalline Fe/V multilayers [6].

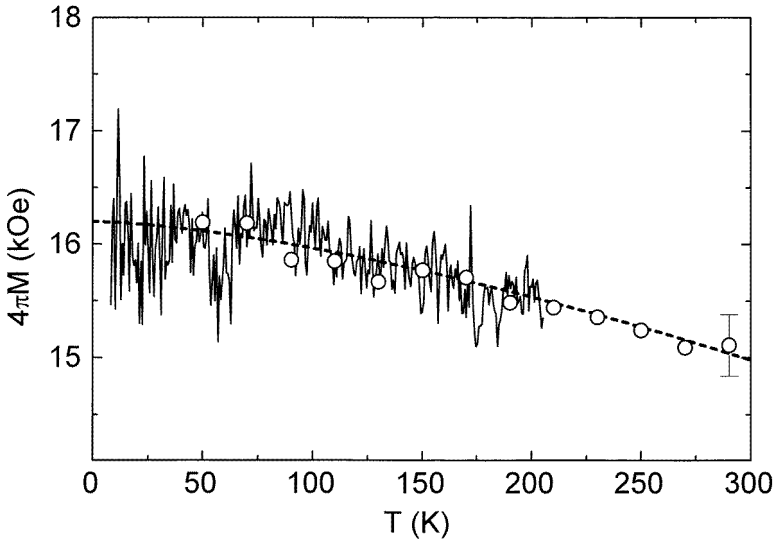

Figure 4. Temperature dependence of the saturation magnetization measured with the static field applied along the [110] direction. Solid line: VSM measurements at $H = 500$ Oe, circles: normalized FMR data at $H = 800$–850 Oe, dashed line: Bloch-law fit.

By using this experimental $M(T)$ we separated the values for the K-constants. The results are shown in figure 5. One may see that $K_{2\perp}(T)$ is nearly constant. Ideally, one expects an $(M(T)/M(0))^3$ dependence for a $K_2(T)$ curve [23] and this has been recently observed in ultrathin Fe/Cu(001) films [24]. Since $M(T)$ is also nearly constant in this temperature range (figure 4), the temperature dependence of $K_{2\perp}$, shown in figure 5, is in reasonable agreement with theory [23]. There are two origins for the out-of-plane $K_{2\perp}$ for tetragonally distorted films such as our samples [4], elastic stress due to lattice mismatch between the constituents and surface anisotropy due to the broken symmetry at the interfaces [10]:

$$K_{2\perp} = K_{2\perp}^{V} + 2K_{2\perp}^{S}/d \tag{3}$$

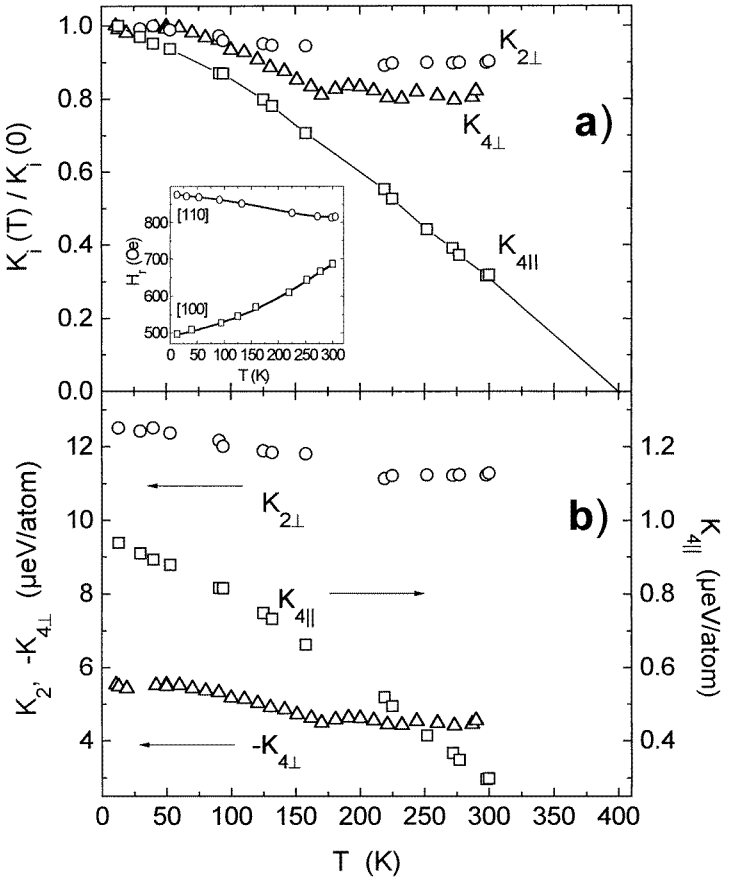

Figure 5. Temperature dependences of the anisotropy constants normalized to unity (a) or in μeV/atom (b). These data were obtained from the H_r measurements at 9.23 GHz. By extrapolating the $K_{4\|}$ values to higher temperatures, an in-plane reorientation from the [100] to [110] should occur at about 400 K. Note the different sign in $K_{4\|}$ and $K_{4\perp}$, discussed in the text. In the inset the temperature dependence of H_r along [110] and [100] is also depicted.

where d is the thickness of the magnetic constituent in each modulation period. By using the values ($\varepsilon_\| = 1.7\times10^{-2}$, $\varepsilon_\perp = -1.3\times10^{-2}$) of the stress tensor for (Fe_n/V_m)/MgO(001) [4], the values of the magnetostriction coefficient λ_{100} and the elastic stiffness constants c_{xy} for bcc Fe [16] one can, in principle, calculate $K^V_{2\perp}$ [9]:

$$K^V_{2\perp} = \tfrac{3}{2}\lambda_{100}(c_{11} - c_{12})(\varepsilon_\perp - \varepsilon_\|). \tag{4}$$

The room-temperature $K^V_{2\perp}$-value obtained this way is $K^V_{2\perp} = -7.87$ μeV/atom. The negative sign indicates that the out-of-plane direction is a hard-magnetization axis. Then, by using the experimental $K_{2\perp}$-value and (3) we may find the room-temperature value for $K^S_{2\perp} = 38.2$ μeV/atom [25]. This value is more than one order of magnitude smaller than the corresponding values for bcc Fe(001)/Ag(001) [17], fcc Fe(001)/Cu(001) bilayers [15], or textured polycrystalline {111} Fe/V multilayers on glass [26] and somewhat smaller than for bulk Fe [27]. Furthermore, the shape anisotropy for our sample is much larger (about

70 μeV/atom). Since the shape anisotropy is the main term of the total free energy density, one should not expect our sample to present any perpendicular anisotropy. In addition, since n is already quite small, it is rather unlikely for any Fe thickness for the Fe_n/V_m series to present perpendicular anisotropy, in agreement with literature reports [6, 26, 28].

In figure 5 the temperature dependence of the out-of-plane $K_{4\perp}$ and the in-plane $K_{4\parallel}$ anisotropy constants are also shown. $K_{4\perp}$ has a negative sign, favouring in-plane orientations; it is weakly dependent on T on a relative scale $K_{4\perp}(300\ \mathrm{K})/K_{4\perp}(10\ \mathrm{K})$ and it is approximately half as large as $K_{2\perp}$. $K_{4\parallel}$ is positive, one order of magnitude smaller than $K_{4\perp}$ and it is strongly temperature dependent. It is interesting to note that on an absolute scale, that is $|K_4(300\ \mathrm{K}) - K_4(10\ \mathrm{K})|$, both $K_{4\parallel}$ and $K_{4\perp}$ have almost the same change with temperature (that is 0.8(2) μeV/atom). This variation between 10 and 300 K is practically the same as that previously found for a 10 nm Fe/MgO(001) sample [29]. (In order to be able to make a direct comparison, let us note here that 1 μeV/atom corresponds to 1.33×10^5 erg cm^{-3}.) From pure symmetry arguments one should not expect a large difference between the in- and out-of-plane fourfold anisotropies, such as the one recorded for our sample. The very small $K_{4\parallel}$ value presented here could be understood by regarding the competition between two anisotropic contributions with different sign. Namely, a $K_{4\parallel}^V$ favouring the [100], as in the bulk, and a $K_{4\parallel}^S$ favouring [110]. This fourfold surface anisotropy has been theoretically predicted [30] and experimentally verified for ultrathin Fe films on various substrates [5, 31, 32]. The fourfold in-plane anisotropy of samples with an Fe thickness near the critical one for the in-plane reorientation was also found to present a much stronger temperature dependence on a relative scale than the corresponding anisotropy constant of thicker Fe films [5], in agreement with our results. By extrapolating $K_{4\parallel}(T)$ to higher temperatures one could expect a change of its sign at about 400 K (figure 5(a)), indicative of a reorientation of the in-plane easy axis from the [100] to the [110] direction. This is also shown in the inset of figure 5 from the reduction in the difference between the [100] and [110] H_r as temperature increases. Such a temperature-dependent in-plane reorientation has been recently observed for W(110)/6 nm Fe(110)/W(110) films [22].

For our series of Fe_n/V_m superlattices a room-temperature isotropic in-plane behaviour has been recorded for two different samples with $n = 3$ and $m = 13$ and 8 ML, respectively [7]. The first of these two samples presented an antiferromagnetic-like coupling as both hysteresis loops and FMR revealed. This coupling resulted in the appearance of several FMR resonance modes. The second sample gave a simpler FMR spectrum. For this sample we have measured the in-plane angular dependence of the resonance field at both 10 K and 300 K. While isotropic at room temperature, the sample presented a slight in-plane anisotropy at low temperatures favouring the [100]. By comparing this result to the corresponding one for the Fe_4/V_4 sample, we find that as n decreases from 4 to 3 ML the temperature-dependent reorientation occurs at lower temperatures. This observation suggests that the [110] direction is the easy axis of the Fe_n/V_m series as n (temperature) decreases (increases).

Our critical thickness of 3 ML is very close to the one (about 4 ML) found by Heinrich *et al* [5] for ultrathin bcc Fe films on various substrates. A recent report [32] for ultrathin Fe(001)/Au(001)/MgO(001) finds the critical thickness to be about 7 ML at room temperature. There is a disagreement between phenomenological approaches for the origin of this fourfold surface anisotropy, regarded as a result of 'ordered crystallographic defects' with an in-plane fourfold symmetry [5] or as an electronic effect related to the specific type of the magnetic–non-magnetic interface [32]. For the Fe(001)/Au(001)/MgO(001) samples a thorough structural analysis has ruled out the first interpretation. Similarly, for samples of high structural quality like ours, as both x-ray diffraction studies [4] and FMR reveal, it

would be rather unlikely to attribute the $K_{4\parallel}^S$ to crystallographic defects.

While the $K_{4\parallel}$ presents such an unusual behaviour, the $K_{4\perp}$ value is in reasonable agreement with the fourfold anisotropy value of thicker Fe films on MgO [29]. This implies that $K_{4\perp}$ does not have a surface term contribution and it merely consists of the volume part.

4. FMR linewidths and sample quality

Frequency-dependent measurements of FMR linewidths allow us to obtain the value of the Gilbert parameter G in the Landau–Lifshitz equation [5]. For a Lorentzian lineshape [5]:

$$\Delta H_{exp}(\omega) = \Delta H_{inhom} + \frac{2}{\sqrt{3}}\frac{G}{\gamma^2 M}\omega. \tag{5}$$

This equation separates the experimentally observed FMR linewidths ΔH_{exp} into two terms: (a) ΔH_{inhom} which is frequency independent and describes the inhomogeneous broadening and (b) ΔH_{hom} originating from the intrinsic damping of the excited spin wave. By plotting ΔH_{exp} as a function of the frequency ω at a given temperature T, one may obtain ΔH_{inhom} and G by a linear fitting via (5). This is depicted in figure 6 at $T = 300$ K. The direction of measurement is the in-plane Fe [110] axis. Such a fitting results in $\Delta H_{inhom} = (17.6 \pm 2)$ Oe and $G = (5.9 \pm 2) \times 10^7$ rad s^{-1}. This G-value is in good agreement with the best literature values especially with those of high-quality Fe whiskers (see table 1).

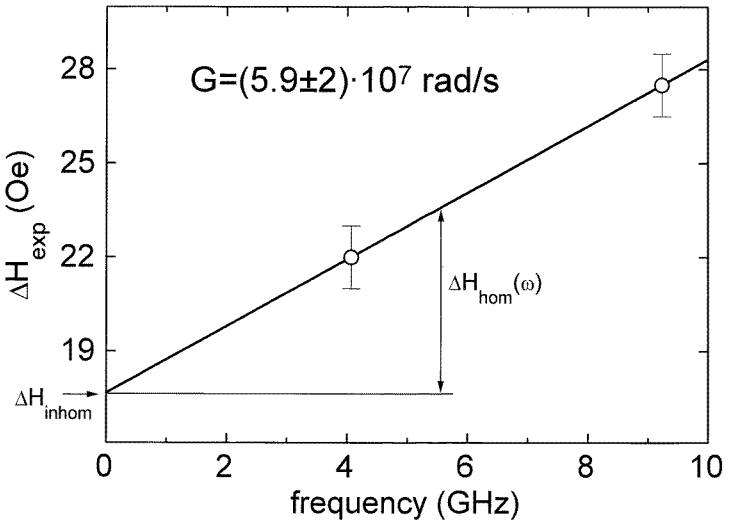

Figure 6. Frequency dependence of the experimental FMR linewidths ΔH_{exp} (circles) measured along the [110] at $T = 300$ K. A linear fitting according to (5) allows a determination of the parameters G and ΔH_{inhom}.

The angular dependences of ΔH_{exp} are shown in figure 7. It is interesting that for the in-plane relative hard axis [110] ΔH_{exp} presents smaller values than for the easy axis [100], in agreement with FMR studies of [39]. Besides this, ΔH_{exp} values are strongly dependent on the orientation of the microwave field: the minimum ΔH_{exp} values are always obtained when both static and microwave fields are lying in the film plane pointing perpendicular to

Table 1. Gilbert parameter G for $(Fe_4/V_4)_{40}$/MgO(001) and other Fe-based samples.

Sample	G ($\times 10^7$ rad s^{-1})	Orientation	Reference
$(Fe_4/V_4)_{40}$/MgO(001)	5.9 ± 2	[110]	This work
Fe whiskers	7	[100]	[33]
Fe whiskers	5.72 ± 0.4	[100]	[27]
Single-crystalline bcc Fe slab, $d = 15$ μm	7.0	[100]	[34]
Single-crystalline bcc Fe (001) discs, $d = 200$, 350 μm	8.0–8.4	[100], [110]	[35]
Single-crystalline bcc Fe slabs, $d = 45$ μm	7.0 ± 0.6	[100]	[36]
1.8–20 nm Fe/GaAs(110)	24	[110]	[37]
Ultrathin bcc Fe/Ag(001)	8.4	[110], [100]	[38]
fcc 16 ML Fe/10.5 ML Cu/Ag(001)	10.8	[110]	[39]
	12.8	[100]	
fcc 3 ML Fe/x ML Cu/3 ML Fe/Cu(001)	21.6–37.8[a]	isotropic	[15]
20 nm bcc Fe/MgO(001)	5.9 ± 0.6	[110], [100]	[14]

[a] Depending on the Cu thickness x.

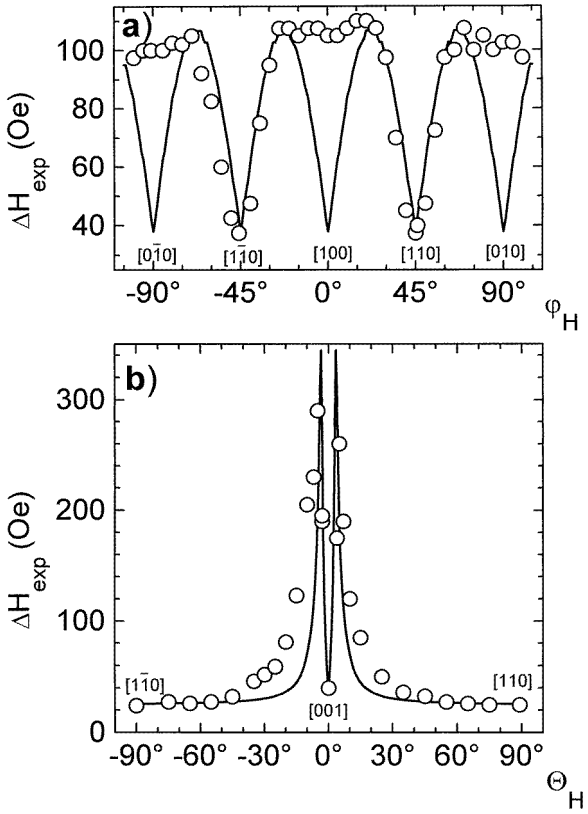

Figure 7. Angular dependences of ΔH_{exp} in (a) and out of (b) plane at 9.23 GHz and 10 K. Circles denote experimental data, while the solid lines are the result of our fit procedure. For the fit parameters see table 2.

each other. This is possibly related to the fact that by this configuration the ellipticity of the precessing magnetization is minimal [40].

According to Chappert *et al* [41] the angular dependence of ΔH_{inhom} may be approximately written in the form:

$$\Delta H_{inhom}(\theta_H) = \Delta\theta_H(\partial H_r/\partial\theta_H) + \Delta\varphi_H(\partial H_r/\partial\varphi_H) + \Delta H_{int}(\partial H_r/\partial H_{int}) \quad (6)$$

where θ_H stands for the angle between the film normal [001] and the external static magnetic field, while φ_H is the angle between the static magnetic field and the in-plane [100] direction. It is obvious from figures 2 and 3 (in particular figure 2(b)), that small variations in $\Delta\theta_H$ and $\Delta\varphi_H$ describing the variation in orientation of the crystallographic axes, create some inhomogeneous broadening. ΔH_{int} is the inhomogeneity of the internal magnetic field throughout the macroscopic sample being measured.

Table 2. Inhomogeneity parameters for $(Fe_4/V_4)_{40}$/MgO(001) and Fe-based samples. The large scatter (±50%) of the fitted parameters ΔH_{int}, $\Delta\theta_H$ and $\Delta\varphi_H$ is due to factors explained in the main text and, additionally, because the model we have used for fitting is an approximate one. Despite this, the experimental values of the FMR linewidths for our films, lines 1–5, are the smallest in table 2, indicating a high magnetic homogeneity.

f (GHz)	T (K)	ΔH_{exp} (G)	Orient.	ΔH_{int} (G)	$\Delta\theta_H$ (°)	$\Delta\varphi_H$ (°)	Reference
9.23	10	38	[110]	13	0.1	5.7	This work
9.23	296	38	[110]	13	0.12	16	This work
4.06	258	22	[110]	2	—	9	This work
4.06	300	19	[110]	2	0.9	9	This work
9.23	300	24	[110][a]				[7]
25	300	100	in-plane				[27]
9	300	39	[110]				[37]
36	300	100	in-plane				[15]
10	300	50	isotropic				[14]
9	293	40	[110][b]				[29]

[a] For 40.9 nm Fe/MgO(001) prepared in the same chamber with the Fe_n/V_m samples.
[b] For 10 nm Fe/MgO(001).

Considering $\Delta\theta_H$, $\Delta\varphi_H$ and ΔH_{int} as fit parameters one may determine their values from the experimental angular dependences of ΔH_{exp}. The results at 9.23 GHz are exhibited in figure 7 (solid lines). $\Delta H_{exp}(\theta_H)$ is obviously described satisfactorily with (6), for example, the minima for $\Delta H_{exp}(\varphi_H)$ in figure 7(b) are found at the position at which H_r in figure 2(b) presents an extreme value (maximum or minimum). The same does not hold for $\Delta H_{exp}(\varphi_H)$ since the $\Delta H_{exp}([100])$ minimum does not coincide at all with the theoretical simulation (figure 7(a)). This discrepancy could possibly have the following origins: first, the anisotropy of the Gilbert parameter [39] which is not included in the present approximate fitting procedure and will be discussed in detail elsewhere; second, a slight misalignment of the axis of the uniaxial anisotropy from [100]. The latter effect could also be responsible for the small systematic deviation of the fitted curve from the experimental data shown in figure 2(a). Nevertheless, we have tried to obtain the inhomogeneity parameters disregarding the above-mentioned incompleteness. The results are listed in table 2. One can see that there is some inconsistency in the fitting parameters at 9.2 and 4 GHz. This occurs because H_r at 4 GHz is comparable to the saturation field and the single-domain approximation, which is a necessary condition for the validity of (6), does not hold. Anyway, the data of table 2 reveal a small value of ΔH_{inhom} (reflecting a good homogeneity of the anisotropy constants) as well as a small spread in the axis of uniaxial anisotropy. This homogeneity is comparable to and even better than the other values given in table 2. For a superlattice with such small m and n numbers, such a large homogeneity in anisotropy parameters comparable

to a well prepared one-element thick film or bulk material [5] is in general not expected. Furthermore, it is interesting to note that the only film in table 2 that presents narrower FMR linewidths than the Fe_4/V_4 superlattice is the 40.9 nm Fe/MgO(001) film prepared in the same chamber and with the same growth technique as the Fe_n/V_m superlattices. Therefore, we have shown that by the UHV-based sputtering method [4] we can grow magnetic films of superior quality, comparable to films produced by the molecular beam epitaxy (MBE) technique.

5. Conclusions

In the present work, the temperature dependencies of the anisotropy constants of single-crystalline $Fe_n/V_m(001)$ superlattices on MgO(001) substrates were studied via FMR. The major anisotropic term was found to be the shape anisotropy favouring an in-plane orientation of the easy-magnetization axis. The out-of-plane twofold anisotropy is a result of a volume part originating from stress and a surface Néel-type term. The first favours in- and the second out-of-plane magnetization. The outcome of this competition is an almost temperature-independent term, favouring out-of-plane orientation, but it is an order of magnitude smaller than the shape anisotropy. Thus a reorientation from in to out of plane for our sample seems to be impossible. The out-of-plane fourfold anisotropy is bulklike. The in-plane fourfold anisotropy is very small, as a result of a competition between a volume term, favouring the [100] and a surface term favouring the [110] direction. It presents an unusually strong temperature dependence, and it possibly leads to an in-plane reorientation phase transition at about 400 K. As the Fe thickness decreases to $n = 3$ ML this reorientation should take place at lower temperatures (about 300 K). A very small step-induced uniaxial anisotropy was revealed, indicating the high sensitivity of the FMR technique in the study of the magnetic anisotropy. Finally, the extremely small FMR-linewidth values revealed films of high homogeneity in agreement with structural studies previously reported [4].

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