Evolution of magnetic anisotropy and spin-reorientation transition in Fe films grown on GaAs(113)A substrates by molecular-beam epitaxy

P. K. Muduli, a) J. Herfort, H.-P. Schönerr, and K. H. Ploog

Paul-Drude-Institut für Festkörperelektronik, Hausvogteiplatz 5-7, Berlin 10117, Germany

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The magnetic properties of Fe films grown on GaAs(113)A substrates by molecular-beam epitaxy are studied using superconducting quantum interference device magnetometry for a wide range of thickness varying from 3.5 monolayers (MLs) to 100 nm (714 MLs). The first signature of ferromagnetism is found at a nominal coverage of about 4 MLs, attributed to a percolation phenomenon, similar to Fe on GaAs(001). The magnetic anisotropy of all samples is found to be a combination of varying strengths of an in-plane uniaxial magnetic anisotropy (UMA) and a four-fold magnetic anisotropy. Samples of thickness $d_{\text{Fe}} \leq 50$ MLs exhibit a dominating UMA with the easy and hard axes along $[332]$ and $[\bar{1}10]$, respectively, whereas samples of thickness $d_{\text{Fe}} \geq 70$ MLs exhibit a dominating four-fold magnetic anisotropy with the easy axes along the in-plane $(0\bar{3}1)$ directions. The reorientation of the easy axis from $[332]$ to the in-plane $(0\bar{3}1)$ axes is found to take place between 50 and 70 MLs, the same thickness range where the relaxation of the layer starts. The effective uniaxial magnetic anisotropy constant $K_{\text{eff}}$ first increases with monolayer coverage up to about 10 MLs and then decreases with the increase in thickness. On the other hand, the effective four-fold anisotropy constant $K_{\text{eff}}$ first increases with monolayer coverage and then saturates close to the bulk value after about 20 MLs. From a comparison of our results with literature a common origin of UMA in Fe films on GaAs(001) and (113)A, i.e., the anisotropy of the bonding of Fe with As and Ga at the interface is anticipated. © 2005 American Institute of Physics. DOI: 10.1063/1.1929852

I. INTRODUCTION

Ferromagnetic metals on semiconductor substrates offer a possibility to realize spin-based devices, i.e., devices that use both the spin and the charge of the electron. 1–3 Fe, having a lattice mismatch of only $-1.4\%$ with GaAs and having a high Curie temperature of 1040 K (for bulk Fe), is a promising candidate for such applications. Because of the small lattice mismatch, the stable bcc phase of Fe grows epitaxially on GaAs(001), GaAs(110), GaAs(113)A, and GaAs(113)A substrates, and numerous studies optimizing its growth conditions and analyzing its interface structure have been reported. 4–13 For efficient spin injection to be achieved, intermixing of Fe with Ga and/or As has to be minimized, which is believed to be detrimental to spin-dependent transmission across the interface. 12 For this reason most of the previous studies of Fe growth were performed on Ga-terminated GaAs(001) templates, at a growth temperature close to room temperature (RT), to avoid interface reactions between Fe and GaAs. 11–13 However, we have recently demonstrated that Fe growth on As-terminated GaAs surfaces is superior to that on Ga-terminated surfaces in removing macroscopic defects on the surface. 13 Moreover, spin injection of Fe into GaAs was achieved due to this optimized growth conditions. 14

Single-crystal Fe films on GaAs also offer an opportunity of studying magnetism at low dimension with controllable magnetic properties. It is well known that ultrathin Fe films on GaAs(001) substrates below a certain thickness exhibit a dominant in-plane uniaxial magnetic anisotropy (UMA) with the easy and hard axes along $[100]$ and $[\bar{1}10]$, respectively, 7,11,15–17 which is unexpected from the crystal symmetry of Fe. The origin of this UMA can not be understood from the combination of common anisotropy energies, such as magnetocrystalline anisotropy, demagnetizing field energy and magnetoelastic coupling energy. Krebs et al. added a uniaxial anisotropy term to the total magnetic anisotropy energy to understand the observed magnetic properties in this system. 18 More recently, Thomas et al. 19 pointed out that for a quantitative modeling of the thickness dependence of the magnetic anisotropy, a strain anisotropy mediated by a magnetoelastic coupling term has to be included though they ruled out magnetoelastic coupling as a cause for the observed UMA. To get further insight into the magnetic properties of the Fe/GaAs system, we have chosen the [113]A orientation of GaAs for three different reasons. First, the reconstructed surface of GaAs(113)A exhibits a step structure, 20–22 which is expected to modify the magnetic properties. Second, such a system could be useful to understand the origin of UMA in the Fe/GaAs system. Third, it is shown in literature that the GaAs(113)A substrate, being a high-index substrate, acts as an excellent template for growing semiconductor nanostructures. 23–28 Thus, it might be suitable for growing magnetic nanostructures of Fe, where the magnetic properties can be manipulated. In addition, for the (113) surface of Fe, none

a)Electronic mail: muduli@pdi-berlin.de
of the bulk easy axes (the ⟨001⟩ directions) lies in plane, thus making it an interesting system for the study of shape anisotropy.

This paper is organized in the following way. In Sec. II experimental details of the preparation of the GaAs template, the growth of Fe, and its basic structural characterizations are discussed. Section III presents the results obtained with ex situ superconducting quantum interference device (SQUID) magnetometry in three subsections dealing with the onset of ferromagnetism, UMA in ultrathin films, and in-plane fourfold anisotropy of thick films in the Fe/GaAs(113)A system. In Sec. IV we discuss an analytical model to explain the observed magnetic properties. The phenomena involving the onset of ferromagnetism and the possible origin of uniaxial magnetic anisotropy in the Fe/GaAs system, in general, are also discussed.

II. EXPERIMENT

Fe films in the thickness range of 3.5 monolayers (MLs) to 100 nm (714 MLs) were epitaxially grown on GaAs(113)A substrates by molecular-beam epitaxy (MBE). The GaAs(113)A templates were prepared in a separate III-V growth chamber under As-rich conditions. The growth temperature for the GaAs buffer layer was 610 °C. All Fe samples presented in this article were grown on an (8 × 1) reconstructed GaAs(113)A surface, obtained by cooling the substrate after the buffer layer growth under As₄ flux to 400 °C. The (8 × 1) reconstruction of GaAs was found to be very stable down to 0 °C. Fe films were grown in an As-free metal chamber connected to the III-V growth chamber through an ultrahigh vacuum Interlock. The growth temperature of Fe was maintained at 0 °C with a typical growth rate of 1.2 ML/min, where we assume a thickness of 0.18 Å for 1-ML coverage similar to the case of Fe on GaAs(001). Fe layers grown under these conditions result in a smooth surface morphology, eliminating macroscopic defects as reported previously. In addition, the low growth temperature prevents the formation of Fe₂As₃ or Fe₂Ga₂₋ₓAsₓ compounds at the interface, which are considered to be detrimental to the spin-injection efficiency. The thickness of the Fe films was calibrated using ex situ x-ray reflectivity (XRR), a technique that can be used to determine the thickness in a wide range from a few to several thousands of angstroms. XRR and high-resolution x-ray diffraction (HRXRD) measurements were done using a PANanalytical X’Pert diffractometer system with a Ge(220) hybrid monochromator using Cu Kα₁ radiation. For the XRR measurements a 0.18° receiving slit was used. Reasonable fits to the experimental XRR curves were obtained by assuming a thin layer of iron oxide of thickness 2.5–2.8 nm, corresponding to approximately 1.5 nm of Fe in agreement with reported values in literature. In Fig. 1 we compare an experimental XRR curve with the simulation of 13-nm-thick Fe and 2.8-nm-thick iron oxide layers on GaAs(113)A. From the simulations of XRR profiles we found the interface roughness of the Fe–GaAs interface to be in the range of 0.3–0.8 nm for all thicknesses studied, indicating a rather smooth interface between Fe and GaAs. The initial growth mode and the magnetic structure of these films were also studied using in situ reflection high-energy electron diffraction (RHEED) with an acceleration voltage of 20 kV. The RHEED pattern from the starting GaAs(113)A substrate shows an (8 × 1) reconstruction with the second-order Laue circle visible, indicating an ordered and smooth surface. The intensity of the specular beam from GaAs decreases drastically after a growth of 1 ML, indicating the formation of disordered islands, and the RHEED pattern was found to be still dominated by the GaAs substrate. After the second monolayer, the RHEED pattern was found to disappear completely and only a diffuse background was seen. The first RHEED pattern from the Fe films was seen at about 4 MLs, suggesting coalescence of islands.

For Fe films thicker than 10 MLs, we observe a streaky RHEED pattern along the [332] and [110] azimuths, characteristic of the unreconstructed bcc Fe lattice. The second-order Laue circle was visible for a RHEED pattern along the [110] azimuth. From high-resolution x-ray diffraction (HRXRD) measurements an epitaxial relationship of Fe[113][332]||GaAs[113][332] was confirmed. Magnetic measurements were performed ex situ, using a commercial Quantum Design SQUID magnetometer, model MPMS-XL. Samples with thicknesses from 3.5 to 140 MLs were studied in the temperature range of 10–300 K. For SQUID measurements a 20-nm-thick Al capping layer was grown on the Fe layers to prevent oxidation. The growth of the capping layer was also performed at 0 °C to avoid diffusion of Al into Fe.

III. RESULTS

A. Onset of ferromagnetism

A plot of the remanent magnetization M₀ versus temperature T for film thicknesses from 4.0 to 30 MLs is shown in Fig. 2. The remanent magnetization M₀ is normalized with respect to its value at 10 K, i.e., M₀/μ₀. These measurements were performed while heating the sample from 10 to 300 K. Before the measurement, the samples were magnetized along the [332] direction up to fields large enough to ensure a complete orientation of the magnetic moments. Since all these measurements were made with the magnetic field applied parallel to the easy axis (see Sec. III B), the remanent magnetization is essentially the same as the saturation magnetization. For samples above 10-ML thickness, the RT remanence changes by less than 10% of its value at 10 K, indicating a high Curie temperature of these samples. Thinner samples below 10 MLs show a drastic change in rema-
nence with temperature. The Curie temperature decreases strongly when decreasing the thickness and reaches 120 K for the 4-ML sample. For the 3.5-ML Fe sample we did not observe any ferromagnetism even at 2 K. This implies that the first signature of ferromagnetic order in our films starts between 3.5 and 4.0 MLs. This is approximately the same thickness for which we observe the first RHEED pattern from the Fe layer. The onset of ferromagnetism in these layers can be ascribed to a percolation phenomenon during the coalescence of Fe islands similar to that observed in Fe on GaAs(001).31,32

The reader may note that the remanent magnetization versus the temperature curve in Fig. 2 for the 4.0-ML sample having a Curie temperature below RT does not exhibit a behavior typical for an ideal two-dimensional (2D) ferromagnet. The remanent magnetization versus the temperature curve does not drop sharply as expected for an ideal 2D ferromagnet and as observed in the magneto-optical Kerr effect (MOKE) data of Bensch et al.31 for Fe/GaAs(001). In fact, susceptibility versus temperature measurements at an ac frequency of 140 Hz and an ac field of 0.04 Oe shows a peak at 107 K with a half width at half maximum (FWHM) of about 100 K. This large FWHM can be attributed to the averaging measurement method of SQUID magnetometry, which measures the magnetization averaged over the whole sample (several mm²) in contrast to the MOKE data of Bensch et al.31 which probes only a small area of the sample. The islanding at the initial stage of growth, in agreement with the RHEED observations, leads to local variations of the Curie temperature within the film and hence to a broadening of the magnetic phase transition. It should be pointed out that a similar behavior was observed in our previous studies of Fe on GaAs(001).32

B. In-plane uniaxial magnetic anisotropy in ultrathin films

Figure 3 shows magnetization curves of the 7-ML-thick Fe film (dFe = 7 MLs) of Fig. 2 at two different temperatures. These measurements were performed with the magnetic field applied parallel to the two major in-plane perpendicular directions, namely, the [332] and [110] directions. The magnetization curves in Fig. 3 and all other magnetization curves in this article are corrected for the diamagnetic contribution of the GaAs substrate. The magnetization M is normalized to the saturation magnetization M₁ at T=10 K, i.e., M₁. As can be seen in Fig. 3, the 7-ML Fe film exhibit a strong UMA with the easy axis along [332]. The hysteresis curve along this direction has a pure rectangular shape with a normalized remanence M₁/M₁ of about one. This indicates that the Fe layer is essentially single domain in nature or with a preferred domain orientation pointing along the easy axis as suggested for Fe/GaAs(001).32 An increase in the coercive field at 10 K is observed compared to 300 K. The magnetization curve along the [110] direction does not show any hysteresis, and the coercive field vanishes. The magnetization curve is completely reversible, indicating a hard axis of magnetization. The reversibility of the magnetization curve is suggestive of a rotation process.8 The in-plane anisotropy field is about 2.5 kOe at 10 K and 1.5 kOe at 300 K. It should be noted that the saturation magnetization for this sample at 300 K decreases to about 10% of its value at 10 K. Moreover, the saturation magnetization for T=0 is close to the value of bulk Fe (1740 emu/cm³), indicating the absence of interfacial Fe–Ga–As compounds.32 It is noteworthy that the UMA shows a hard axis along the [001] direction similar to the case of Fe on GaAs(001).

C. Four-fold magnetic anisotropy in thick films

We study the evolution of magnetic anisotropy by performing a series of magnetization measurements at RT for Fe film thicknesses varying between 7 MLs and 100 nm. In
The saturation magnetization of this 100-nm-thick Fe film calculated from the coherent rotation model for 500 Oe. This saturating field of 500 Oe is close to the coercive field. Fig. 4 shows an abrupt transition at a very low field of about 2 Oe, followed by a gradual increase to saturation at approximately 300 Oe. In the intermediate direction of bulk Fe, the magnetization curve of the 50-ML sample is similar to that of the 7-ML sample. Hence the 50-ML sample still exhibits a rectangular shape with the normalized remanence of about one. It should be mentioned that the measurement results for the 50- to 70-ML Fe.

To investigate the reorientation phenomenon in more detail, we performed HRXRD measurements on samples having thicknesses varying from 50 MLs to 100 nm. Figures 5(a) and 5(b) show, respectively, the x-ray reciprocal space maps of the diffracted intensity of the two samples of 50- and 70-ML thickness. The measurements were done near the asymmetric (004) reflection of GaAs (in grazing incidence geometry), since the counterpart of the symmetric GaAs(113) reflection is forbidden for Fe. The reciprocal lattice units (rlu) are λ/2d, where λ is the wavelength of Cu Ka1 radiation and d is the lattice plane spacing of the corresponding reflection. The GaAs(004) reflection is located near Qx = 0.233 and Qy = 0.492. For the 50-ML sample, the layer peak is located at ∆Qx = 0 and ∆Qy = 0.0146, relative to the GaAs substrate where ∆Qx,y = Qx,y,0 - Qx,y,layer. The parallel mismatch is nearly zero and the layer is completely coherently strained.
strained. In contrast, the layer peak for the 70-ML sample is at \( \Delta Q_z = 0.0016 \) and \( \Delta Q_y = 0.0129 \), relative to the GaAs substrate. The parallel mismatch in this case is about 6880 ppm, indicating that the layer is partly relaxed. Pendellösung fringes are visible for the sample of 50-MLs, indicating the good crystal quality of the film. Thus, the relaxation of the layer starts between 50 and 70 MLs, the same thickness range where the reorientation of the easy axes takes place. This indicates that the relaxation of the layer and reorientation of the easy axes of magnetization are apparently correlated. However, we cannot rule out that the reorientation phenomenon and the relaxation of the layer just happen to occur in the same thickness range and are not necessarily related. Thomas et al.\(^\text{33}\) recently reported that Fe films on GaAs(001) below 60 MLs (exhibiting a dominating UMA) are strained and those above 60 MLs (exhibiting a dominating four-fold anisotropy) are relaxed. Hence, the anisotropic strain relaxation in thicker films had to be taken into account for a quantitative modeling of the magnetic anisotropy as a function of Fe film thickness.

It is important to note that the easy axis of magnetization in thick films \((d_{Fe} \approx 70\) MLs\) is found to be along the in-plane \((03\bar{1})\) axes, which are not the easy axes of magnetization of bulk Fe. In fact, one of the easy axis of bulk Fe, namely the \([001]\) direction, lies out of plane at an angle of 25.24° to the surface normal and towards the [332] direction. However, we found no evidence of a perpendicular easy axis effect, details of which are discussed elsewhere.\(^{34}\)

\[ \text{IV. DISCUSSION} \]

The magnetization of Fe/GaAs is often understood by means of a simple rotational mechanism.\(^{15,8,35,36}\) In order to understand the observed magnetic properties of Fe on GaAs(113)A, we use a phenomenological expansion of the free-energy density. The free-energy density of a single domain of magnetization \( M \) in a magnetic field \( H \), with a combination of cubic magnetocrystalline and uniaxial anisotropy can be written as

\[
E = K_1 [\alpha_1^2 \alpha_2^2 + \alpha_2^2 \alpha_3^2 + \alpha_3^2 \alpha_1^2] + K_2 [\alpha_1^2 \alpha_2^2 \alpha_3^2] + K_u (M \cdot u/M)^2 + 2 \pi (M \cdot n)^2 - M \cdot H, \tag{1}
\]

where \( K_1 \) and \( K_2 \) are the first two cubic anisotropy constants, \( K_u \) is the uniaxial anisotropy constant, and the last two terms refer to the demagnetization energy and Zeeman energy, respectively. Here, \( \alpha_1, \alpha_2, \) and \( \alpha_3 \) are direction cosines of \( M \) relative to the cubic crystalline axes. The unit vectors \( u \) and \( n \) represent the direction of the uniaxial anisotropy and the surface normal, respectively.

For an in-plane magnetic field, the anisotropy energy density, \( E_{\text{IPMA}} \) of the \((113)\) surface of Fe can be derived from Eq. (1) by using the direction cosines of the \((113)\) surface symmetry and neglecting the last term, which represents the demagnetization energy. Hence,

\[
E_{\text{IPMA}} = \frac{(K_1^{\text{eff}}/484)[89 + 16 \cos 2\theta_M + 48 \cos 4\theta_M]}{\sin^3 \theta_M - MH \cos \phi} + K_u^{\text{eff}} \sin^2 \theta_M \tag{2}
\]

Here \( \theta_M \) is the angle of magnetization \( M \) with respect to the unit vector \( u \), which is taken along the [332] direction, the observed easy axis of the in-plane UMA. \( \phi \) is the angle of magnetization \( M \) with respect to the magnetic field \( H \). The term involving \( K_2 \) is neglected in the above expression, since for bulk Fe and at RT \( K_2 \) is smaller than \( K_1 \) by two order of magnitudes.\(^{37}\) In Eq. (2), we have introduced the new constants \( K_1^{\text{eff}} \) and \( K_u^{\text{eff}} \) to include the thickness dependence

\[
K_1^{\text{eff}} = K_1 + K_1^{\text{int}}/d_{Fe}, \tag{3}
\]

\[
K_u^{\text{eff}} = K_u + K_u^{\text{int}}/d_{Fe},
\]

where \( K_1^{\text{int}}, K_u^{\text{int}}, K_1, \) and \( K_u \) describe the in-plane volume and interface four-fold and uniaxial anisotropy. Here, \( d_{Fe} \) is the thickness of the Fe film. \( K_1^{\text{int}} \) is assumed to comprise the contribution of the interface between the magnetic film and the substrate as well as the interface to the Al capping layer, respectively.

The observed magnetic properties can be qualitatively understood from Eq. (2) above. In Fig. 6 we show a polar plot of Eq. (2) for different anisotropy ratios \( r = K_u^{\text{eff}}/K_1^{\text{eff}} \) at zero applied field. The major in-plane directions are also shown, with zero degree along the [332] direction.

![Polar plot of in-plane anisotropy energy density \( E_{\text{IPMA}} \) at zero field of the Fe (113) surface according to Eq. (2) for different anisotropy ratio \( r = K_u^{\text{eff}}/K_1^{\text{eff}} \). The major in-plane directions are also shown, with zero degree along the [332] direction.](image2.png)
ratio (i.e., \( r \geq 1.0 \)), UMA dominates and the minimum now lies along the \([332]\) direction, which becomes the easy axis. This explains the behavior of the magnetization curves observed for the Fe films of and below 50 MLs in Figs. 4 and 3. Thus, a continuous variation of the anisotropy ratios with thickness can explain the observed magnetic properties.

To get more insight in the magnetic anisotropy constants in dependence of the thickness, we use the hard-axis magnetization curves in the UMA regime to separate the four-fold and the uniaxial magnetic anisotropy. This is done by assuming a reversible rotation along the hard axis \([\bar{1}10]\) similar to that adopted by Brockmann et al.\(^{38}\) for Fe on GaAs(001). An analytical expression for the inverse magnetization loop \(H(m)\) can be obtained by minimizing the energy given by Eq. (2) for a magnetic field along \([\bar{1}10]\) (\(\phi = \pi/2 - \theta_m\)). The resulting expression for \(H(m)\) is given by

\[
H(m) = K_{eff}^{110}(384m^3 - 280m)/121M_s + 2K_u m/M_s,
\]

where \(m = M/M_s\) is the normalized component of the magnetization along \([\bar{1}10]\). This is valid in the range of the coherent rotation of the magnetization, i.e., for \(-1 < m < 1\). For samples with \(d_{Fe} \leq 50\) MLs, the hard-axis magnetization curve along the \([\bar{1}10]\) direction was fitted in the range of \(-1 < m < 1\) to yield \(K_{eff}^{110}\) and \(K_{eff}^{\bar{1}10}\). In Fig. 7 we show plots of \(K_u\) and \(K_{eff}^{110}\) versus film thickness \(d_{Fe}\) at (a) 10 K and (b) 300 K obtained from the fits. For all examined temperatures (10–300 K), \(K_u\) first increases with the monolayer coverage until about 10 MLs and then decreases with further increase of the monolayer coverage. On the other hand, \(K_{eff}^{110}\) first increases with the monolayer coverage and then saturates to the bulk value after about \(\sim 20\) MLs. The decrease of \(K_u\) below 10 MLs is correlated to the structural properties of the Fe films and can be qualitatively understood in terms of the nucleation of Fe islands during the initial stages of growth similar to Fe on GaAs(001).\(^{39,32}\) To fully understand the thickness dependence of \(K_{eff}^{110}\) after 10 MLs a more extended analysis of the dependence of the effective anisotropy constants on thickness and specially of the separation of volume and interface anisotropy constants is required. Nevertheless, from a comparison of our results with literature\(^{38,39}\) for Fe/GaAs(001), we believe that this thickness dependence arises from the interfacial nature of the UMA, i.e., \(K_{eff}^{110}\) is a pure interface related term. This is supported by the fact that the fitting of the hard-axis inverse magnetization loop for Fe films below \(\sim 5\) MLs yields \(K_{eff} = 0\); i.e., the magnetization curve in the range of \(-1 < m < 1\) is almost linear in this thickness range.

Finally, we address the question on the origin of UMA in the Fe/GaAs(113)A system. A particularly interesting possible origin of UMA in Fe on GaAs(113)A could be the surface structure of the reconstructed GaAs(113)A surface, which exhibit a surface corrugation with step edges parallel to the \([332]\) direction.\(^{20-22}\) It is known that Fe films grown on stepped surfaces exhibit a uniaxial magnetic anisotropy.\(^{30,41}\) For Fe on stepped W(001), Chen and Erskine\(^{40}\) reported a step-induced uniaxial anisotropy with the easy axis perpendicular to the steps, whereas for Fe on stepped Ag(001) Kawakami et al.\(^{31}\) reported an easy axis parallel to the step edges. However, we did not observe any split loop in the magnetization curves like in the cases of Fe on stepped metallic surfaces\(^{40,41}\) or in other systems, e.g., Co on vicinal Cu(001).\(^{42}\) Moreover, we observe qualitatively similar magnetic properties in Fe/GaAs(113)A compared to Fe/GaAs(001). So we conclude that these reconstruction-induced steps are probably not responsible for the observed UMA. We observe a hard axis along \([\bar{1}10]\), which is also a hard axis for Fe on GaAs(001). The easy axis of UMA in Fe on GaAs(113)A found along the \([332]\) direction can be considered as a projection of the \([\bar{1}10]\) axis onto the (113) surface. Hence, the atomic configuration along both of the \([332]\) and \([\bar{1}10]\) directions should be similar. We thus believe that there is a common origin for UMA in the Fe/GaAs(113)A and the Fe/GaAs(001) systems. For Fe/GaAs(001), the surface reconstruction of GaAs was shown to be not responsible for the observed UMA.\(^{7,43}\) Since the uniaxial anisotropy in our samples is found to be originating from the Fe/GaAs interface, we anticipate that the UMA in the Fe/GaAs system is determined by the anisotropy of the bonding between Fe and Ga at the Fe–GaAs interface. This is supported by the results of Kneidel et al.,\(^{7}\) where the directional nature of the Fe–As bonding was suggested as an origin of the observed UMA, from a combined scanning tunneling microscopy and Magneto-optic Kerr effect study. A detailed study of the atomic configuration at the Fe–GaAs interface could provide more insight into the microscopic origin of the observed UMA. It should be mentioned that strain anisotropy mediated by magnetoelastic coupling was ruled out as a cause for the observed UMA in thinner Fe films on GaAs(001).\(^{19}\)

V. CONCLUSIONS

We have investigated the structural and magnetic properties of MBE-grown Fe films on GaAs(113)A substrates using RHEED, HRXRD, and SQUID magnetometry. The
SQUID measurements of these Fe films display qualitatively similar magnetic properties compared to Fe on GaAs(001) substrates. The onset of ferromagnetism is observed at a nominal thickness of about 4 MLs, which can be attributed to a percolation phenomenon during the coalescence of Fe islands. The magnetic properties of these films display an intermixing of the four-fold and the uniaxial magnetic anisotropy. For Fe film thicknesses of and below 50 MLs we found a dominating UMA with the easy axis along the in-plane (031) directions. The reorientation of the easy axis from the k-fold anisotropy with the easy axes along the in-plane normalities of and above 70 MLs, we found a dominating four-fold anisotropy with the easy axes along the in-plane (031) directions. The reorientation of the easy axis from the [332] direction to the (031) directions is found to take place between 50 and 70 MLs, the same thickness range where the relaxation of the layer starts. The observed magnetic properties are found to be in good agreement with a simple rotational model of magnetization reversal. By comparing our results with that of Fe on GaAs(001) substrates, a common origin of UMA in Fe/GaAs(113)/A and Fe/GaAs(001) is suggested. The anisotropy of bonding between Fe and Ga at the Fe–GaAs interface is anticipated as a possible origin of the observed UMA.

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