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Magnetic and structural properties of epitaxial Fe thin films on GaAs(0 0 1) and interfaces

M. Doi[a], B.Roldan Cuenya[a], W. Keune[a,]*, T. Schmitte[b], A. Nefedov[b], H. Zabel[b], D. Spoddig[c], R. Meckenstock[c], J. Pelzl[c]

[a] *Laboratorium fur Angewandte Physik, Gerhard-Mercator-Universität, Duisburg, D-47048 Duisburg, Germany*
[b] *Experimentalphysik/Festkörperphysik, Ruhr-Universität-Bochum, D-44780 Bochum, Germany*
[c] *AG Festkörperspektroskopie, Ruhr-Universität-Bochum, D-44780 Bochum, Germany*

Abstract

Fe(0 0 1) thin films (70 Å) with ^{57}Fe(7.2 Å) tracer layers at the interface were epitaxially grown on GaAs(4×6) surfaces. Magneto-optic Kerr effect and Ferromagnetic resonance measurements indicate a dominant 2-fold in-plane magnetic anisotropy (easy axis along [1 1 0]) superimposed to a 4-fold anisotropy, and small coercivity ($\sim$10 Oe). Mössbauer (CEMS) measurements indicate no magnetic "dead layer" and an average Fe moment of ~ 1.7–$2\,\mu_B$ at the Fe/GaAs interface. © 2002 Elsevier Science B.V. All rights reserved.

Keywords: Iron/gallium arsenide; Magneto-optic Kerr effect; Ferromagnetic resonance; Mössbauer effect; Reflection high-energy electron diffraction

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Magnetic films epitaxially grown on semiconducting substrates have a high potential for technical applications (see for instance Ref. [1]). For this purpose knowledge of the state of the interface is important. It was demonstrated recently by magnetometry that low-temperature growth of Fe on (Ga-terminated) GaAs(0 0 1)(4×6) surfaces inhibits magnetic "dead layer" formation [2,3], or creates half-magnetization phases due to atomic intermixing at the interface [4]. In our present study Mössbauer spectroscopy (CEMS) on thin interfacial ^{57}Fe-isotope probe layers was employed, combined with Reflection high-energy electron diffraction (RHEED), Magneto-optic Kerr effect (MOKE) and Ferromagnetic resonance (FMR).

An MBE system (base pressure 9×10^{-11} mbar) was used to prepare the samples. The substrates were cleaned by Ar$^+$ sputtering (0.5 keV) at 600°C for 30 min. After this, in-situ RHEED images of the substrate (Fig. 1(a)) revealed the pseudo (4×6) surface reconstruction, characteristic of the clean flat Ga-terminated GaAs(0 0 1) surface [2]. Then we deposited 7.2 Å (5 monolayers, ML) of 95% enriched ^{57}Fe isotope, followed by 70 Å of natural Fe (deposition pressure: $< 2 \times 10^{-10}$ mbar; rate: 0.03 Å/s). The substrate temperature was 40–50°C during deposition. The samples were coated by 40 Å of Sn for protection.

After deposition of 5 ML of ^{57}Fe and above, the spotty fundamental reflections in the RHEED patterns (Fig. 1(b) and (c)) are typical for epitaxial BCC–Fe(0 0 1) 3D island growth. From the separation of the reflections in reciprocal space the relative Fe in-plane atomic distance during growth has been determined (Fig. 1(d)). After an initial strong increase which we ascribe to initial intermixing of the interface during island growth, the in-plane atomic distance above $\sim$5 ML thickness remains $\sim$1.3% larger than that of GaAs. This agrees with Ref. [2] and with the lattice mismatch between bulk BCC Fe and GaAs of 1.38%. We conclude that the epitaxial Fe films are not significantly strained in-plane.

The observed CEM spectrum (Fig. 2) was least-squares fitted with two subspectra: a sextet with sharp lines and a magnetic hyperfine (hf) field of 32.8 T due to "bulk-like" BCC Fe, and a broad sextet with a distribution of hf fields, $P(B_{\text{hf}})$, ascribed to a concentra-

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*Corresponding author. Tel.: +49-0-203-379-2387; fax: +49-0-203-379-3601.

E-mail address: keune@uni-duisburg.de (W. Keune).

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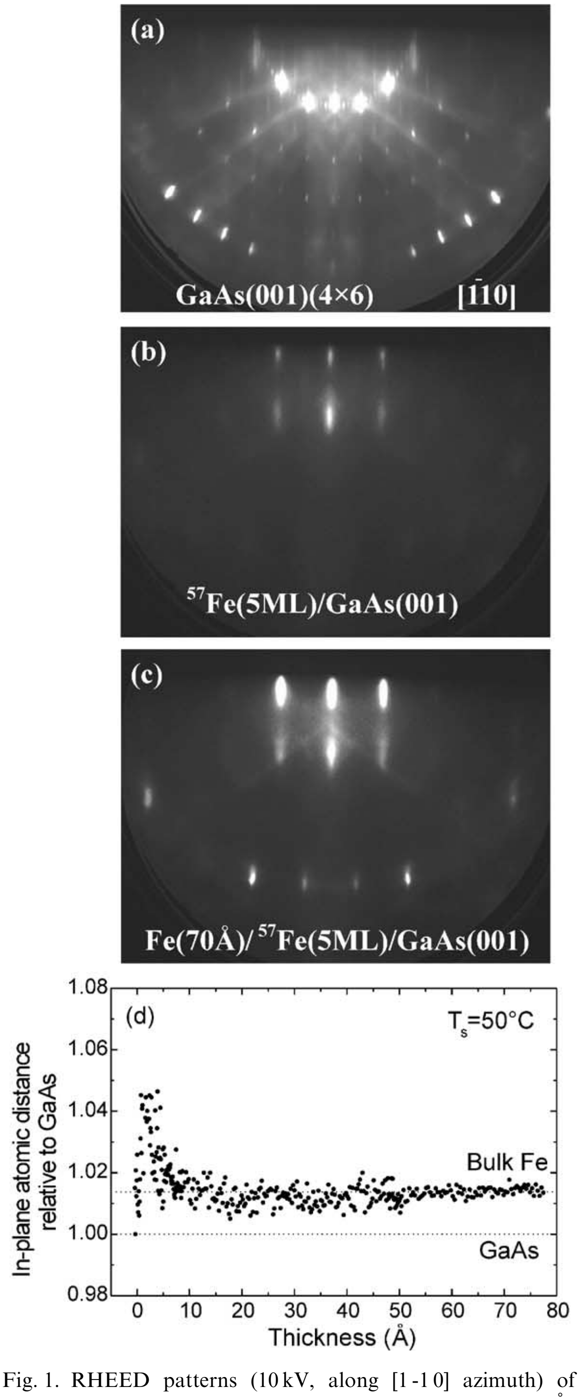

Fig. 1. RHEED patterns (10 kV, along [1 -1 0] azimuth) of clean GaAs(0 0 1)(4 × 6) substrate (a), covered by 5 ML (7.2 Å) of ^{57}Fe (b), followed by 70 Å of natural Fe (c), in-plane atomic distance (relative to GaAs) versus Fe film thickness (d).

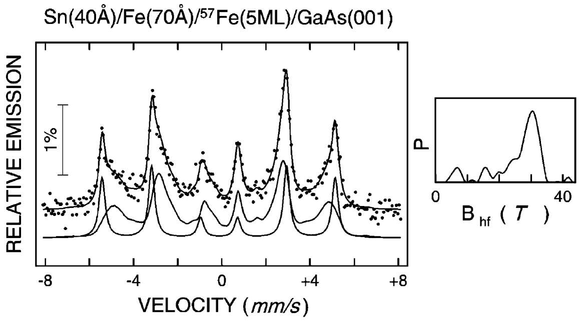

Fig. 2. Mössbauer spectrum (CEMS) and hyperfine magnetic field distribution (on the right).

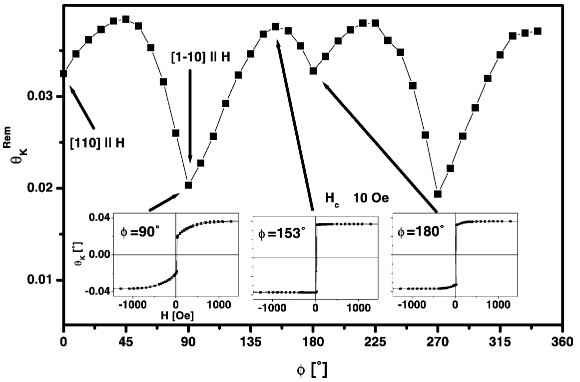

Fig. 3. Kerr-rotation angle θ_K^{Rem} measured at remanence versus the in-plane rotation angle ϕ. Inserts: typical Kerr hysteresis loops at specific ϕ values. At $\phi = 0°$ (90°) B_{ext} is along the [1 1 0] direction ([1 -1 0] direction) of the substrate.

tion gradient (very likely of Ga atoms in an Fe-rich Fe–Ga alloy) at the intermixed Fe/GaAs interface. Since a peak at 0 T is not observed in $P(B_{\text{hf}})$, a magnetic dead layer does not exist at the interface. Moreover, the most probable and average hf field in the distribution are 30.5 and 26.0 T, respectively. By using the usual conversion factor of 15 T/μ_B we deduce corresponding Fe atomic moments of ~ 2 and $\sim 1.7\,\mu_B$, respectively. Thus the interface contains high Fe moments, and large hf fields, similar to those in ferromagnetic Fe–Ga alloys [6].

Magnetic hysteresis curves were measured using longitudinal MOKE with different in-plane rotational angles ϕ between the in-plane applied field H and the in-plane crystallographic axes of the substrate (Fig. 3, inserts). The remanence plotted versus the angle ϕ (Fig. 3) indicates the superposition of a dominant in-plane 2-fold (uniaxial) magnetic anisotropy and a weaker in-plane 4-fold anisotropy. The 2-fold anisotropy has easy axes along the [1 1 0] direction of the substrate (hard axes along [1 -1 0]). The 4-fold anisotropy has easy directions at $\phi \approx 45°$, 135°, 225° and 315°. The origin of the 4-fold anisotropy is the crystalline anisotropy of BCC–Fe, while the uniaxial anisotropy is due to interface anisotropy [5]. The small coercive field of ~ 10 Oe indicates good crystalline film quality.

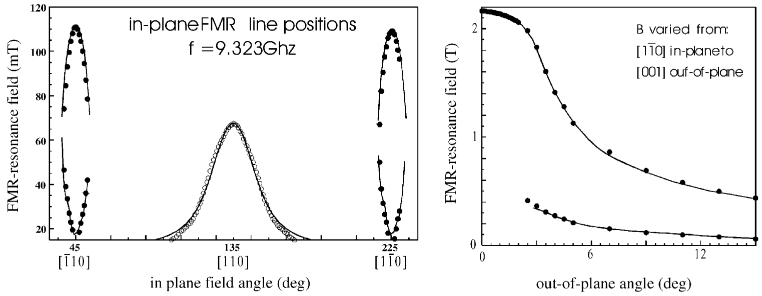

Fig. 4. Angular dependence of FMR line position: dependence on external field angle in the Fe(0 0 1) film plane (left) and dependence on out-of-plane external field angle in the (1 1 0) plane (right).

Our angle dependent FMR investigations yield the following magnetic anisotropy fields [7]: $B_{\mathrm{eff}} = \mu_0 M - 2K_s/tM = 1.9\,\mathrm{T}$, $K_1/M = 19.4\,\mathrm{mT}$, $K_{1s}/M = 13\,\mathrm{mT}$, $K_u/M = 11.2\,\mathrm{mT}$, and g-factor = 2.09. ($M$ is saturation magnetization, t Fe film thickness, K_s surface anisotropy, K_1 in-plane crystalline anisotropy, K_{1s} out-of-plane crystalline anisotropy due to tetragonal distortion, K_u in-plane uniaxial anisotropy.) In order to obtain these parameters FMR was performed with the external field B_{ext} oriented either in the (0 0 1) plane (Fe film surface), the (1 -1 0) plane, or in the (-1 -1 0) plane. In Fig. 4 (left) the measured FMR-line position is plotted versus the in-plane B_{ext} angle in the (0 0 1)-Fe plane. One can clearly distinguish the influence of the 4-fold in-plane crystalline anisotropy and the 2-fold uniaxial in-plane anisotropy. The uniaxial hard axis is oriented along [-1 1 0] and [1 -1 0], and the uniaxial easy axis is along [1 1 0], in agreement with our MOKE results. A pecularity of FMR at 9.325 GHz on epitaxial Fe films is the observation of two lines for certain B_{ext} orientations (for details see Ref. [7]). In Fig. 4 (right) the FMR line position is plotted versus the out-of-plane B_{ext} angle. At about 3° (i.e. close to the film normal direction) the position of both lines is extremely sensitive to a tetragonal distortion of Fe described by K_{1s}.

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