Growth, structure and morphology of epitaxial Fe(0 0 1) films on GaAs(0 0 1)c(4 × 4)

T Ashraf1, C Gusenbauer2,4, J Stangl2,4, G Hesser3 and R Koch2,4

1 National Institute of Lasers and Optronics, Islamabad, Pakistan
2 Institut für Halbleiter- und Festkörperphysik, Johannes Kepler Universität Linz, Altenberger Straße 69, 4040 Linz, Austria
3 Zentrum für Oberflächen- und Nanoanalytik, Johannes Kepler Universität, Altenbergerstrasse 69, A-4040 Linz, Austria
E-mail: reinhold.koch@jku.at

Received 25 August 2014, revised 20 November 2014
Accepted for publication 25 November 2014
Published 24 December 2014

Abstract
We studied epitaxy, growth, structure and morphology of thin Fe(0 0 1) films on the As-rich GaAs(0 0 1)c(4 × 4) surface, deposited by molecular beam epitaxy at growth temperatures between room temperature and 250° C. Electron and x-ray diffraction (XRD) techniques evidence epitaxial growth with Fe(0 0 1)[1 0 0]∥GaAs(0 0 1)[1 0 0]. The residual strain derived from the XRD results is consistent with recent stress measurements. Cross-sectional transmission electron microscopy reveals an abrupt interface for room-temperature films and the formation of a ∼10 nm thick crystalline Fe–Ga–As intermediate layer at 250° C. The dependence of the surface morphology on growth temperature and annealing evidences a kinetic roughening of the Fe surface at growth temperatures of 100–200° C due to the presence of step-edge barriers.

Keywords: Fe, GaAs(0 0 1), epitaxy, interdiffusion, growth mode

(Some figures may appear in colour only in the online journal)

1. Introduction
The deposition of highly spin-polarized ferromagnetic layers on semiconductor substrates is essential for the fabrication of efficient spin-injectors for spintronic [1] and magnetoelectronic devices [2]. With a high Curie temperature of 770° C for bulk Fe and a spin polarization of the conduction electrons of 42% [3, 4], Fe epilayers on GaAs(0 0 1) have become a model system for spin-injection into semiconductors [5–11]. Electrical spin injection at 4.5 K into a GaAs quantum well from an Fe/AlGaAs Schottky contact yielded a spin polarization as high as 32% [8].

High-quality epitaxial bcc-Fe(0 0 1) films are growing on GaAs(0 0 1) as reviewed in [12]. Epitaxial Fe(0 0 1) films have been obtained irrespective of the reconstruction or composition (As-rich or Ga-rich) of the GaAs(0 0 1) surface at growth temperatures between −150 and +300° C [13–26]. For Fe[1 0 0]∥GaAs[1 0 0], where four Fe atoms sit in one GaAs(0 0 1) unit cell, thus occupying different lattice sites, the misfit is calculated to 1.38% (δGaAs = 0.5653 nm, 2aFe = 0.5732 nm). Misfit strain of 1.38% corresponds to stress of 2.9 GPa, which has been observed to develop for room temperature (RT) growth of Fe/GaAs(0 0 1)c(4 × 4) up to a film thickness of about 3 nm indicating pseudomorphic growth where the stress is dominated by the misfit strain; at higher thicknesses part of the strain is relieved by misfit dislocations [25]. At growth temperatures above room temperature (RT) interdiffusion between Fe and GaAs sets in, leading to an increasing degradation of the interface [11, 13, 19, 25, 27]. At still low growth temperatures (<100° C) Fe, Ga and As coexist in an extended interface region on interstitial and substitutional sites [13, 19]. At more elevated growth temperatures or upon annealing (<250° C)
interface reactions lead to the formation of crystalline phases of Fe–Ga–As (e.g. Fe₃As, Fe₃Ga, or Fe₃Ga₂₋ₓAsₓ) [11, 28–31]. Furthermore, segregation of As to the surface of the Fe film has been reported [13, 17, 28]. Improvement of the spin-injection efficiency is expected from optimizing the quality of growth as well as of the film/substrate interface with respect to both its abruptness and type of the participating atoms [32]. For example, in case of Fe/AlGaAs hybrid structures a decrease in interface roughness increases the spin-injection efficiency [33].

Here we report on the epitaxial growth of Fe(001) films on the As-rich GaAs(001)-(4×4) surface at different growth temperatures. We investigated the Fe films in-situ by reflection-high- and low-energy electron diffraction (RHEED and LEED) as well as ex-situ by x-ray diffraction (XRD), cross-sectional transmission electron microscopy (TEM) and atomic force microscopy (AFM). The obtained results on film structure and morphology of bulk and surface provide detailed insight into growth mode and interdiffusion of Fe/GaAs(0 0 1).

2. Experimental

The experiments were performed in a multichamber molecular beam epitaxy (MBE) system consisting of separate interconnected growth chambers for III/V semiconductor and metal MBE. The Fe films were deposited onto c(4×4) reconstructed GaAs(001) substrates mounted on our cantilever-beam device. They were prepared in the III/V growth chamber by following routine: oxide desorption at 615°C, low-temperature buffer at 480°C, high-temperature buffer growth at 590°C, annealing at 605°C, controlled cooling to 380°C at constant As₄ partial pressure of ≈1×10⁻⁶ mbar to form the c(4×4) reconstruction as confirmed by RHEED (figure 1, left), further cooling at reduced As₄ pressure, transfer to the As-free metal chamber without breaking ultrahigh vacuum conditions. The c(4×4) reconstruction is maintained during sample transfer into the metal growth chamber as confirmed by LEED (figure 1, right). The primitive unit cell of the c(4×4) reconstruction is indicated by the dashed square in the LEED pattern; notice the resulting fourfold periodicity along (100) and a twofold one along (110). The Fe films were deposited by e-beam evaporation from an Fe rod within 45 min after the sample transfer; the pressure during deposition was 5×10⁻¹⁰ mbar, the deposition rate was 0.15 nm min⁻¹ as controlled by a quartz crystal microbalance (QCM) calibrated by a QCM in substrate position. A cantilever beam device [34] was used for measuring the stress of the Fe films in real time during growth (stress results are presented in [25]). The XRD and TEM investigations were performed ex-situ with a Seifert XRD3003 (Cu Kα = 0.15418 nm) and a JEOL-2011 TEM, respectively. The AFM investigations were performed with nanoSurf AFM, in the constant-force contact mode.

3. Results

Figure 2 shows LEED results of 15 nm thick Fe films on GaAs(00 1)-c(4×4). The measurements were performed in-situ at room temperature (RT) immediately after the preparation of the Fe films at different growth temperatures (T₀). The observed diffraction patterns exhibit a fourfold symmetry with distinct spots, thus evidencing the successful growth of epitaxial Fe(001) films. By increasing film temperature no major difference in the LEED patterns is detected.

Figure 3 shows XRD spectra of the 15 nm thick Fe films deposited at different growth temperatures, measured ex-situ at RT. In addition to the strong GaAs(004) substrate peak located at 33.057° only one peak located, e.g. for the RT film, at 32.0° is observed. It originates from the Fe layer and corresponds to the Fe(002) peak, in accordance with the LEED results. By using ε || = −(c₁₁/2c₁₂)ε⊥ and the RT lattice constant of Fe (a₀ = 0.2866 nm) a lateral compressive strain of 1.2% is calculated from the diffraction angle of the RT film. With increasing growth temperature, the Fe(002) peak shifts more and more towards the GaAs(004) peak indicating an increasing relief of the compressive misfit strain. For the 250°C film the (002) peak has shifted to 32.38° corresponding
to a vertical Fe–Fe distance of 0.288 nm and indicating an almost strain-relaxed Fe film(0 0 1). Moreover, for growth temperatures above 100° C there is also an increasing intensity at lower diffraction angles. At a growth temperature of 250° C a second peak at 31.5° is clearly visible (marked by arrow in figure 3), corresponding to a perpendicular lattice spacing of 0.295 nm and indicating the formation of a crystalline Fe–Ga–As phase (see discussion section).

Further insight is obtained from cross-sectional TEM images of two Fe films deposited at RT and 250° C, which provide important in-depth information on the interface structure between Fe and GaAs(0 0 1). The RT film (figure 4(a)) exhibits a sharp interface, where the transition from the GaAs to the Fe lattice proceeds within 1–2 atomic layers (unfortunately the Fe lattice is not resolved due to the limited spatial resolution of our TEM). The sample grown at 250° C (figures 4(b) and (c)), on the other hand, has experienced a drastic degradation of interface. As discussed in detail in [25], a broad intermixed Fe–Ga–As layer between the Fe film and the GaAs(0 0 1) substrate has formed. Its thickness is approximately 10 nm and the TEM images of figure 4(c) showing the interface region at higher magnification, reveal a crystalline structure (see inset of figure 4(c), left). The crystalline structure of the intermediate layer is not homogeneous but exhibits lattice distances perpendicular to the growth direction that are both, smaller and larger, compared to the GaAs lattice.

All samples have been investigated ex-situ by AFM immediately after removing them from the UHV chamber, in order to minimize oxide formation on the Fe films. However, since own thin Fe films after years of storage in air remain metallic, we estimate the effect of oxidation as small contributing negligibly (<0.5 nm) to a change in surface morphology. Figure 5 displays AFM images of the 15 nm-thick Fe films deposited at the indicated growth temperatures. Obviously the surface morphology of the Fe(0 0 1) films depends strongly on the growth temperature. The RT film is very flat. Its overall surface morphology on a scale of 10 µm as determined by AFM is very similar to that of uncoated GaAs(0 0 1)-c(4 × 4) substrates, indicating that the imaged surface morphology of RT films is dominated mainly by the substrate topography. The root-mean-square (rms) roughness measured on the extended terraces of the RT Fe film is smaller than the expected error owing to partial oxidation of film and substrate in air (<0.5 nm), respectively. With increasing growth temperature the surface corrugation...
Figure 5. 500 × 500 nm² AFM micrographs of the GaAs(0 0 1)-c(4 × 4) buffer layer and of Fe films on GaAs(0 0 1)-c(4 × 4) prepared at different growth temperatures: RT, 100, 150, 200 and 250 °C as well as 200 °C annealed for 10 min at 200 °C; the vertical scale is identical for all linescans and indicated in top, left.

The LEED investigations of Fe/GaAs(0 0 1)-c(4 × 4) prepared at growth temperatures between RT and 250 °C yield sharp diffraction patterns with fourfold symmetry (figure 2) evidencing—in combination with the stress results of [25]—that epitaxial bcc Fe films are growing on the GaAs(0 0 1)-c(4 × 4) surface with Fe(0 0 1)[1 0 0] || GaAs(0 0 1)[1 0 0]. Compared to the LEED pattern of the GaAs(0 0 1)-c(4 × 4) buffer layer (figure 1, right) the peaks of the Fe films are significantly broader indicating considerably smaller terrace widths for the epitaxial Fe(0 0 1) films. Cross-sectional TEM (figure 4) reveals the formation of an abrupt interface for the RT films, while it is degraded at a growth temperature of 250 °C indicating strong intermixing. In the latter case a ∼10 nm thick crystalline intermediate layer has developed at the boundary between the GaAs(0 0 1) substrate and the 250 °C Fe film. Its structure is not uniform but contains regions with lattice spacings in the growth direction that are smaller and larger compared to the GaAs lattice constant indicating the formation of crystalline Fe–Ga–As phases. For instance, the left TEM image figure 4(c), displays an ordered area with a lattice distance of ∼0.25 nm, which indicates the local formation of Fe₂As [30]. On the right side of figure 4(c) a region with more extended lattice spacings is shown.

The effects of intermixing can be recognized also in the XRD spectra. At growth temperatures of 200 °C the intensity of the Fe peak is already significantly reduced. The XRD spectra of the 250 °C film include a distinct additional diffraction peak at ∼31.5° corresponding to lattice spacing of 0.295 nm. This is in good agreement with the lattice constants of ∼0.29 nm reported for various bcc Fe₁₋ₓGaₓ alloys in [35]. Also the diffraction angle of Fe₂Ga₂₋ₓAsₓ of 30.7° lies in that range but is not observed in our XRD spectra [30]. Obviously the composition of the intermediate layer varies locally and favors the growth of a strain-relaxed epitaxial Fe(0 0 1) layer on top of it in accordance with our
Table 1. Comparison of the vertical lattice spacing of the Fe(001) films $d_{\text{XRD}}^{\text{Fe}(001)}$ calculated via Bragg’s law from the XRD data with $d_{\text{XRD}}^{\text{Fe}(001)}$ derived from the stress results of [25].

| $T_0$ [°C] | $d_{\text{XRD}}^{\text{Fe}(001)}$ [nm] | $d_{\text{XRD}}^{\text{Fe}(001)}$ [nm] | $\epsilon_{\text{\sigma}}$ [GPa] | $\epsilon_{\text{G}}^{\text{Fe}(001)}$ [GPa] | $\sigma^{\text{Fe}(001)}$ [%] | $\epsilon_{\text{\sigma}}^{\text{Fe}(001)}$ [%] | $\epsilon_{\text{\sigma}}^{\text{Fe}(001)}$ [%] | $d_{\text{XRD}}^{\text{Fe}(001)}$ [nm] |
|------------|---------------------------------|---------------------------------|-----------------|---------------------------------|-----------------|---------------------------------|-----------------|---------------------------------|
| RT         | 0.2909                          | -1.21                           | -1.67           | -0.80                           | 2893            | -1.67                           | -0.80           | 2893                            |
| 100        | 0.2898                          | -0.90                           | -1.67           | -0.80                           | 2890            | -1.67                           | -0.80           | 2890                            |
| 150        | 0.2893                          | -0.76                           | -1.06           | -0.51                           | 2881            | -1.06                           | -0.51           | 2881                            |
| 200        | 0.2885                          | -0.54                           | -0.80           | -0.38                           | 2874            | -0.80                           | -0.38           | 2874                            |

recent stress measurements [25]. First hints for interdiffusion of Fe, Ga and As can be found already at much lower growth temperatures. For instance, a weak diffraction signal at 31.7° is observed already in the XRD spectrum of the 100 °C Fe film. Interdiffusion at already very low temperatures may explain the discrepancy of previous spin-injection efficiency measurements of FeGaAs(001) [5–11]. Finally we want to remark that due to non-uniform strain distribution—as evidenced by our recent stress investigation [25]—as well as interdiffusion, the peak width in the XRD spectra is larger than estimated by the Scherrer equation for 15 nm thick films.

Table 1 compares the vertical lattice spacing $d_{\text{XRD}}^{\text{Fe}(001)}$ of the Fe(001) films calculated via Bragg’s law from the XRD data with $d_{\text{XRD}}^{\text{Fe}(001)}$ derived from the stress results of [25]. The lateral strain $\epsilon_{\text{\sigma}}^{\text{Fe}(001)}$ is obtained from the stress $\sigma$ measured at $T_0$ (from table 1) via the stress-strain relation:

$$\epsilon_{\text{\sigma}} = \frac{c_{11}}{c_{11} + c_{12} - 2c_{12}} \sigma$$  \hspace{1cm} (1)$$

The $c_{ij}$ are the respective elastic constants of bcc-Fe taken from [36] ($c_{11} = 230$ GPa, $c_{12} = 134$ GPa). From $\epsilon_{\text{\sigma}}^{\text{Fe}(001)}$ the lateral Fe spacing at $T_0$ can be calculated by using the equilibrium lattice constants of Fe at the respective growth temperatures (thermal expansion coefficient: $\alpha_{\text{Fe}} = 11.8 \cdot 10^{-6}/$°C). Upon cooling the samples to RT the GaAs substrate contracts ($\alpha_{\text{GaAs}} = 5.73 \cdot 10^{-6}/$°C). Since the Fe films adhere to the substrate, the lateral Fe–Fe spacing is reduced accordingly. With the resulting RT Fe–Fe spacing the respective in-plane strain is calculated and converted to vertical strain by $\epsilon_{\text{\sigma}} = -2c_{12}/c_{11}$ in order to obtain $d_{\text{XRD}}^{\text{Fe}(001)}$. Table 1 reveals a very good agreement between $T_0$ $d_{\text{XRD}}^{\text{Fe}(001)}$ measured at RT and $d_{\text{XRD}}^{\text{Fe}(001)}$ obtained from the stress data (experimental error: <10%).

Growth and surface morphology of FeGaAs(001) has been studied intensively in the previous years [12]. A direct comparison of these results, however, is difficult because of the many different surface reconstructions of GaAs(001) that, moreover, are either Ga- or As-rich. On Ga-rich GaAs(001) surfaces, typically prepared by Ar-sputtering and annealing in the absence of an As background pressure, surface roughening was observed. For instance, on Ga-rich GaAs(001)–c(8×2) Chambers et al [13] report the formation of small Fe clusters in 4 monolayers (ML) thick Fe films deposited at 175 °C; from their (Auger) electron diffraction and photoelectron spectroscopy data (LEED, AED, XPS) they infer a Volmer–Weber growth mode. A ‘more three-dimensional (3D)’ growth mode is also reported for Fe deposited up to 3 ML at 175 °C onto Ga-rich GaAs(001)–c(4×6) [37]. On Ga-rich GaAs(001)–c(2×6) Godde et al [38] observed a surface roughness of 0.7 nm (~5 ML Fe) in 10 ML thick Fe films; they explained their scanning tunneling microscopy (STM) results by a kinetic roughening due to the step-edge barriers of Fe(001). As-rich substrates, on the other hand, typically are prepared by standard III/V-semiconductor molecular beam epitaxy. Takeshita et al [39] investigated RT growth of the first monolayer of Fe on the As-rich GaAs(001)–c(4×4) with STM; they found that Fe nucleates in form of small 2D clusters whose size is related with the underlying c(4×4) reconstruction. On the As-rich GaAs(001)–c(4×4) surface Thibado et al [15] demonstrated with STM that 2D Fe islands nucleate exclusively on top of the As dimer rows; the islands preferentially coalesce along the rows and persist at least up to an Fe thickness of 5 nm. On As-rich GaAs(001)–c(4×4) the Fe islands appear substantially larger for a given coverage confirming that the initial GaAs reconstruction profoundly influences Fe adsorption and island nucleation [16]. On the latter two surface reconstructions there was no evidence for a kinetic roughening up to an Fe thickness of 5 nm.

Our AFM investigations provide insight into the growth mode of Fe on GaAs(001)–c(4×4). Our results evidence a corrugated surface for ~15 nm thick Fe(001) films (figure 5) above RT. The corrugation increases with the growth temperature up to 200 °C, whereas at even higher temperatures a flattening sets in. This behavior is consistent with a kinetic roughening mechanism for FeGaAs(001)–c(4×4), analogous to the homoepitaxial growth of FeGaAs(001) [40] or the heteroepitaxial growth of Fe/MgO(001) [41]. A comparison with these systems is reasonable, because in all cases Fe is growing on Fe(001) terraces once the substrate surface is completely covered by Fe. Then the large step-edge (Ehrlich–Schwoebel) barriers of the Fe(001) surface suppress the step-down diffusion of atoms, which gives rise to surface roughening. Due to the increase of the residence time of atoms on an existing 2D island, new 2D islands can nucleate also on top of 2D existing islands, thus converting them into 3D islands before they coalesce with neighboring 2D islands.

Kinetic 3D islanding rather than thermodynamic Volmer–Weber growth for FeGaAs(001)–c(4×4) is corroborated by the fact, that annealing of the 200 °C film leads to a flattening and not to a rougher surface (figure 5) because the step-edge barriers are overcome at elevated temperatures. Our AFM results therefore are in support of a kinetic surface roughening during Fe deposition at 100–200 °C in agreement with previous studies [38, 42].

However, the growth of Fe on GaAs(001)–c(4×4) certainly is more complex than on Fe(001) or MgO(001). Contrary to these two systems, As atoms are diffusing from the FeGaAs(001) interface into the Fe film and eventually segregating to its surface. Chambers et al [13] detected a strong As signal even in 65 ML thick Fe films (~10 nm). According to density functional theory calculations of Erwin et al [43] a complete As adlayer lowers the surface energy.
of Fe(001) by $\sim1.5$ eV. Since As is is well known as a efficient surfactant, e.g. on Ge/Si(001) [44], it may affect the magnitude of the step-edge barriers of Fe(001) and thereby change the growth mode of Fe from a kinetic 3D island mode to a layer-by-layer mode. Since the As coverage on the Fe(001) films of Fe/GaAs(001)-c(4×4) depends on the Fe film thickness, the surfactant effect of As may be larger for thinner Fe films. This may explain why Kneedler et al. [16] did not observe kinetic roughening on As-rich GaAs(001)-c(4×4) up to a film thickness of 5 nm. On Ga-rich substrates, on the other hand, Gester et al. [42] and Godde et al. [38] found evidence for regular arrays of pyramid-like surface structures for Fe thicknesses well below 15 nm. STM studies on both types of substrates, revealed that the Fe covers the GaAs(001) substrates already at thicknesses of a few monolayers [16, 38]. Obviously, despite considerable intermixing Fe wets GaAs(001), which opposes Volmer–Weber growth and provides further support for a kinetically induced roughening.

5. Conclusions

In summary, we studied epitaxy, growth, structure and morphology of thin Fe(001) films on the As-rich GaAs(001)-c(4×4) deposited by molecular beam epitaxy at growth temperatures between room temperature and 250 °C. Electron and x-ray diffraction techniques evidence epitaxial growth with Fe(001)[1 0 0]∥GaAs(001)[1 0 0]. The residual strain derived from the XRD results is consistent with recent stress measurements. Cross-sectional transmission electron microscopy reveals an abrupt interface for room-temperature films and the formation of a $\sim10$ nm thick crystalline Fe–Ga–As intermediate layer at 250 °C. The dependence of the surface morphology on growth temperature and annealing suggests a kinetic roughening of the Fe surface at growth temperatures of 100–200 °C due to the presence of step-edge barriers.

Acknowledgments

We thank the Austrian Fonds zur Förderung der Wissenschaftlichen Forschung (Project P20650), the German BMBF under the research program NanoQUIT (Contract No. 01BM463) and Higher Education Commission (HEC) of Pakistan for financial support. We also thank National Institute of Lasers and Optronics (NILOP) Islamabad, Pakistan for their cooperation.

References

[1] Datta S and Das B 1990 Appl. Phys. Lett. 56 665
[2] Ney A, Pampuch C, Koch R and Ploog K H 2003 Nature 425 485
[3] Soulun R J 1998 Science 282 86
[4] Tobin J G, Yu S W, Morton S A, Waddill G D, Thompson J D W, Neal J R, Pengenberg M and Shen T H 2010 Surf. Sci. 604 1342
[5] Zha H J, Ramsteiner M, Kostial H, Wassermeier M, Sch H-P and Ploog K H 2001 Phys. Rev. Lett. 87 016601
[6] Hanbicki A T, Jonker B T, Itskos G, Kioseoglou G and Petrov A 2002 Appl. Phys. Lett. 80 1240
[7] Strand J, Schultz B D, Isakovic A F, Palmstr C J and Crowell P A 2003 Phys. Rev. Lett. 91 036602
[8] Hanbicki A T, van Erve O M J, Magno R, Kioseoglou G, Li C H, Jonker B T, Itskos G, Mallory R and Petrov A 2003 Phys. Rev. Lett. 82 4092
[9] Adelmann C, Xie J Q, Palmstr C J, Strand J, Lou X, Wang J and Crowell P A 2005 J. Vac. Sci. Technol. B 23 1747
[10] Adelmann C, Lou X, Strand J, Palmstr C J and Crowell P A 2005 Phys. Rev. B 71 121301
[11] Schultz B D, Marom N, Naveh D, Lou X, Adelmann C, Strand J, Crowell P A, Kronik L and Palmstr C J 2009 Phys. Rev. B 80 201309
[12] Wastlbauer G and Bland J A C 2005 Adv. Phys. 54 137
[13] Chambers S A, Xu F, Chen H W, Vitomirov I M, Anderson S B and Weaver J H 1986 Phys. Rev. B 34 6605
[14] Krebs J J, Jonker B T and Prinz G A 1987 Appl. Phys. Lett. 61 2596
[15] Thibado P M, Kneedler E D, Jonker B T, Bennett B R, Shanabrook B V and Whitman L J 1996 Phys. Rev. B 53 R10481
[16] Kneedler E M, Jonker B T, Wang R J, Shanabrook B V and Whitman L J 1997 Phys. Rev. B 56 8163
[17] Monchesky T L, Heinrich B, Urban R, Myrtle K, Klaus M and Kirschenker J 1999 Phys. Rev. B 60 10242
[18] Sch H-P, Nötzol R, Ma W and Ploog K H 2001 J. Appl. Phys. 89 169
[19] Wedler G, Wassermann B, Nötzol R and Koch R 2001 Appl. Phys. Lett. 78 1270
[20] Hirohata A, Steinmueller S J, Cho W S, Xu Y B, Guertler C M, Wastlbauer G, Bland J A C and Holmes S N 2002 Phys. Rev. B 66 035330
[21] Moosb R, Bensch F, Dumm M and Bayreuther G 2002 J. Appl. Phys. 91 8757
[22] Herfort J, Braun W, Trampert A, Sch H-P and Ploog K H 2004 Surf. Sci. 534 120
[23] Ashraf T, Gusenbauer C, Stangl J, Hesser G, Wegscheider M and Koch R 2005 Appl. Surf. Sci. 240 49
[24] Fleet L R, Yoshida K, Kobayashi H, Kaneko Y, Matsuzaka S, Ohno Y, Ohno H, Honda S, Inoue J and Hirohata A 2013 Phys. Rev. B 87 044401
[25] Mitir S, Sanyal B, Ishedon C and Johansson B 2003 Phys. Rev. B 67 155421
[26] Thompson J D W, Neal J R, Shen T H, Morton S A, Tobin J G, Waddill G D, Matthew J A D, Greig D and Hopkinson M 2008 J. Appl. Phys. 104 024516
[27] Rahmouni M, Eymery J P and Denanot M F 1997 J. Magn. Magn. Mater. 175 219
[28] Lépine B et al 1998 J. Appl. Phys. 83 3077
[29] Lallaiizion C, Lépine B, Ababou S, Schussler A, Quémerais A, Ohno Y, Ohno H, Honda S, Inoue J and Hirohata A 2013 Phys. Rev. B 87 044401
[30] Hülser B, Scheffler M and Krätzer P 2009 Phys. Rev. Lett. 103 046802
[31] Zega T J, Hanbicki A T, Erwin S C, Zewi I, Kioseoglou G, Li C H, Jonker B T and Stroud R M 2006 Phys. Rev. Lett. 96 196101
[32] Weber M, Koch R and Rieder K-H 1994 Phys. Rev. Lett. 73 1166
[33] Matsushita M, Matsushima Y and Ono F 2010 Physica B 405 1154
[34] Adams J J, Agosta D S, Leisuré R G and Ledbetter H 2006 J. Appl. Phys. 100 113530
[35] Jonker B T, Prinz G A and Idzerda Y U 1991 J. Vac. Sci. Technol. B 9 2437
[36] Godde C, Noor S, Urban C and Köhler U 2008 Surf. Sci. 602 3343
[39] Takeshita H, Akinaga H, Ehinger M, Suzuki Y, Ando K and Tanaka K 1995 Japan. J. Appl. Phys. 34 1119
[40] Stroscio J A, Pierce D T, Stiles M D, Zangwill A and Sander L M 1995 Phys. Rev. Lett. 75 4246
[41] Thürmer K, Koch R, Weber M and Rieder K H 1995 Phys. Rev. Lett. 75 1767
[42] Gester M, Daboo C, Gray S J and Bland J A C 1997 J. Magn. Magn. Mater. 165 242
[43] Erwin S C, Lee S-H and Scheffler M 2002 Phys. Rev. B 65 205422
[44] Copel M, Reuter M C, Kaxiras E and Tromp R M 1989 Phys. Rev. Lett. 63 632