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Ferromagnetic metal/semiconductor heterostructures for magneto-electronic devices

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Abstract

We have successfully grown single crystal Fe films on InAs(100) and InAs(graded)/GaAs(100) substrates using molecular beam epitaxy. In situ magneto-optical Kerr effect (MOKE) and ex situ alternating gradient field magnetometry (AGFM) measurements show that the films have well defined magnetic properties, and *I–V* measurements in the temperature range 2.5–304 K show that Fe forms an ohmic contact on InAs. This demonstrates that Fe/InAs is a very promising system for use in future magneto-electronic devices as it has *both* favorable magnetic and electrical properties. We also show that with careful substrate preparation and suitable growth conditions Fe/GaAs films do not exhibit a magnetically 'dead' layer at the interface. A spin-polarized field effect transistor based on Fe/InAs/GaAs has been proposed, which could operate using either an external electric field or an external magnetic field. © 2000 Elsevier Science S.A. All rights reserved.

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1. Introduction

Over the past three decades solid state electronics has developed dramatically, but with very little attention devoted to incorporating magnetic materials into integrated electronic devices. However, an exciting new field of electronics, magnetoelectronics [1], has attracted much attention recently, which is based on the fact that electrons have spin as well as charge. Future magneto-electronic devices, in which the spin of the electron is controlled, are expected to find applications based on hybrid semiconductor structures. A spin-polarized field effect transistor (spin FET) has already been proposed [2], which is based on the injection of spin-polarized electrons from a ferromagnetic source into the semiconductor.

The development of magnetoelectronics stems essentially from the successful epitaxial growth of high quality single crystal ferromagnetic metal (FM) films on oriented semiconductor substrates, where the magnetic properties can be controlled. The most extensively studied system to date is Fe/GaAs. Epitaxial growth has been achieved by several groups [3–6], due in part to the fact that the lattice constant of bcc Fe ($a_o = 2.866$ Å) is almost exactly half that of GaAs ($a_o = 5.654$ Å), as shown in Fig. 1. However, as Fe forms a rectifying contact on GaAs, the Schottky barrier (~ 0.8 eV for Fe/GaAs [7]) prevents efficient spin-injection from the FM pads to the semiconductor substrates. The fabrication of ever smaller devices leads to higher current densities, which in turn need low resistance ohmic contacts to reduce thermal dissipation. Metals on narrow gap semiconductors, such as InAs which has a direct band gap as small as 0.36 eV at 300 K (Fig. 1), form low resistance contacts [7]. InAs also has a higher low-field mobility than GaAs and InP, which makes it an excellent candidate for high speed field effect transistors [8]. The crystal structure of InAs is very similar to that of GaAs, namely a zincblende structure composed of two nested face-centered-cubic cells [9]. Though the lattice mismatch of Fe and InAs ($a_o = 6.058$ Å) of 5.4% is much larger than that of Fe/GaAs (1.3%) (as shown in Fig. 1), it is possible that bcc Fe may stabilize on InAs through a lattice relaxation process. It is well known that in semicon-

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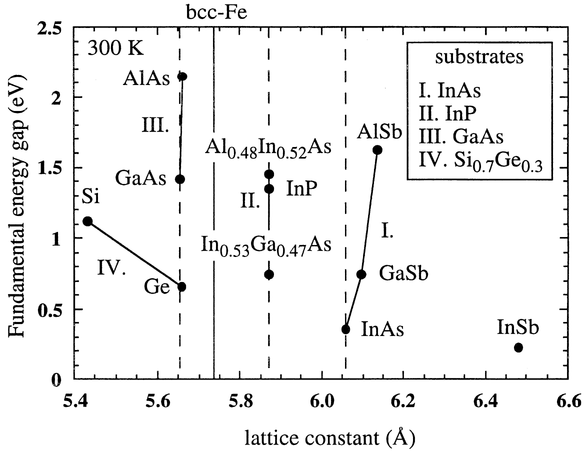

Fig. 1. The fundamental energy gap against lattice constant for several common III–V compounds and silicon–germanium. The lattice constant (×2) is shown for bcc Fe.

ductor growth, such as InAs on GaAs [10], high quality epitaxy can be achieved despite the large lattice mismatch of two materials (7.1%). Fig. 1 also shows several classes of III–V materials, as well as the strained silicon–germanium system, that are closely lattice matched to InAs, InP, and GaAs. In this paper we demonstrate that single crystal bcc Fe can be grown epitaxially on InAs(100) and InAs(graded)/GaAs substrates, and that the electron tunneling across Fe/InAs interface is energy barrier-free. We also show that with a well prepared substrate, Fe/GaAs films do not exhibit a magnetically 'dead' layer at the interface, which would be detrimental to the spin-dependent electron tunneling.

2. Experiments

This study was carried out in a 'multi-technique' molecular beam epitaxy (MBE) system, which includes in situ magneto-optical Kerr effect (MOKE) and Brillouin light scattering (BLS) to probe the static and dynamic magnetic properties of samples, and scanning tunneling microscopy (STM), low energy electron diffraction (LEED), reflection high energy electron diffraction (RHEED), and Auger spectroscopy to provide structural, morphological and compositional information. The Fe films were grown at a rate of approximately one monolayer (ML) per minute using an e-beam evaporator. The pressure during growth was around $7–8 \times 10^{-10}$ mbar. The deposition rate was monitored by a quartz microbalance which was calibrated using RHEED oscillations of Fe on an Ag(100) single crystal. The MOKE hysteresis loops were collected in the longitudinal geometry with an intensity stabilized HeNe laser [11]. Current–voltage (I–V) characterization was performed ex situ in the 304–2.5 K temperature range using a Keithley 236 source-measure unit. The magnetization was measured ex situ using alternating field gradient magnetometry (AGFM) with sensitivity up to 10^{-6} emu.

3. Fe / GaAs(100)

The substrates used in this study are As capped GaAs(001) prepared in another UHV chamber. A buffer layer (~0.5 μm) of homoepitaxial GaAs was grown on the commercial wafer to provide the smoothest possible GaAs surface. The As capping layer began to desorb at around 340°C and the substrate was further annealed to 550°C for 1 h to obtain a clean and ordered surface. Fe films were grown at room temperature rather than higher temperature (~175°C) to reduce the intermixing of Fe with Ga or As at the interface. However, higher temperature growth usually results in better epitaxy [3,5]. The LEED picture (Fig. 2a) of the substrate shows a very clear $p(4 \times 6)$ reconstruction, typical for Ga rich surfaces [12]. This clear and sharp LEED pattern for the reconstructed surface indicates that the GaAs substrate surface is very flat and well crystallized. Auger measurements show that the substrate is free of O and C after As desorption. The LEED patterns were monitored as Fe was deposited. Clear LEED patterns were observed after the deposition of 5 ML. The LEED pattern of a 140 ML Fe film is shown in Fig. 2b, which demonstrated that Fe grows epitaxially on GaAs(001) at room temperature with the epitaxial relationship Fe(001)⟨100⟩||GaAs(001)⟨100⟩.

The evolution of the ferromagnetic phase has been studied in detail using in situ MOKE measurements. We found that the magnetic phase proceeds via three phases; a non-magnetic phase for the first three and a half monolay-

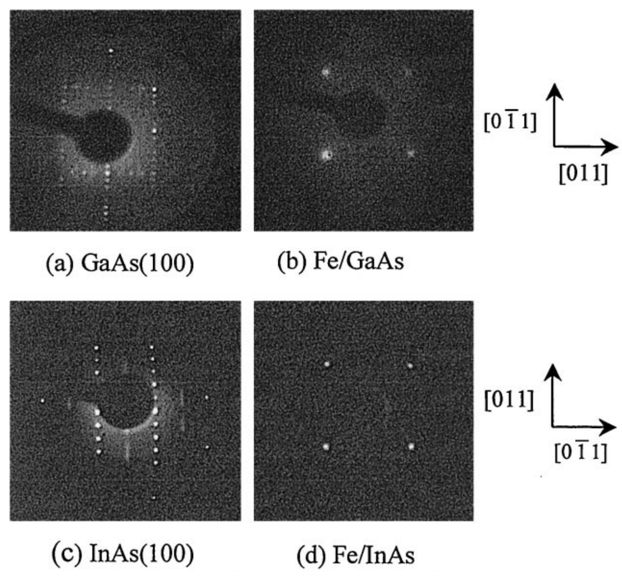

Fig. 2. LEED patterns of (a) GaAs(100) 135 eV, (b) Fe(140 ML)/GaAs(100) 136 eV; (c) InAs(100) substrate 68 eV, and (d) Fe(50 ML)/InAs(100) 136 eV.

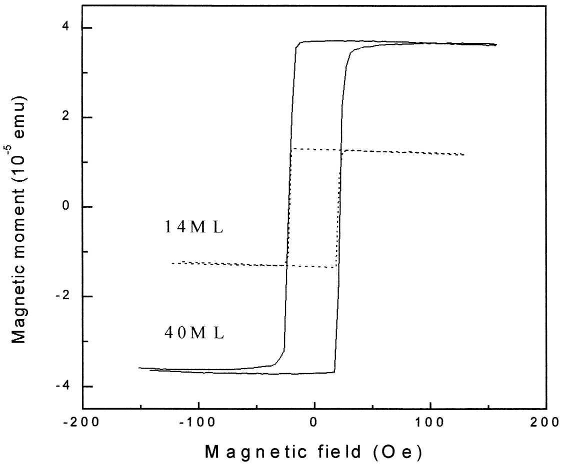

Fig. 3. Magnetization hysteresis loops of (a) Fe(14 ML)/GaAs(001) and (b) Fe(40 ML)/GaAs(001) measured using AGFM with the magnetic field.

ers, a short-range-ordered superparamagnetic phase, and a ferromagnetic phase above about five monolayers. Detailed results are presented in Ref. [13]. The in situ MOKE results indicate that the magnetization does not vary strongly with thickness after the onset of the ferromagnetism. The magnetizations of two Fe/GaAs samples have also been measured using AGFM. Fig. 3 shows the magnetization hysteresis loops of two samples; (a) Au(20 ML)/Fe(14 ML)/GaAs(001) and (b) Au(20 ML)/Fe(40 ML)/GaAs(001). The total magnetic moment from the 40 ML of Fe is about 2.8 times bigger than that from the 14 ML, which is in proportion to their thicknesses. The magnetization of the films is $1.6 \pm 0.2 \times 10^3$ emu/cm^3, only slightly smaller than that of the bulk bcc Fe(1.71×10^3 emu/cm^3). The AGFM measurements further show that the magnetization is approximately thickness-independent and the Fe films have a bulk-like moment.

4. Fe / InAs(100)

The InAs(100) substrates used are commercial wafers. Before loading into the UHV system, the substrates were cleaned using a combination of oxygen plasma etching and wet etching (HCl: $H_2O = 1{:}4$). The LEED picture of the substrate after annealing at 510°C for half an hour is shown in Fig. 2c. The clearly 4×2 reconstructed LEED pattern indicates that the cleaning procedure produces a well-ordered surface. Auger measurements show that after annealing the substrate is free of O, but contains a small amount of C. Fe films were grown at 175°C, which is expected to produce better epitaxial growth, but may introduce intermixing at the interface. The LEED patterns were monitored as Fe was deposited. After approximately 5 ML of deposition, faint LEED spots from the Fe film appear. The LEED spots become clearer and sharper with increasing coverage up to 25 ML and then retain a similar shape for higher coverages. A typical LEED pattern is shown in Fig. 2d after 50 ML of deposition, which demonstrates that single crystal bcc Fe films have been stabilized on InAs(100) $- 4 \times 2$. The epitaxial relationship is Fe(100) $\langle 001 \rangle \| $InAs(100)$\langle 001 \rangle$, i.e., the same as that for the Fe/GaAs system.

The magnetic properties were studied with in situ MOKE measurements. A clear magnetic signal was detected after 4 ML of deposition. The amplitude and the remanent ratio of the MOKE-loops increase with increasing coverage. Above about 15 ML of Fe, square loops were observed when the field is applied along the $\langle 001 \rangle$ directions, similar to those in Fig. 4a, which shows MOKE-hysteresis loops for an Fe(50 ML)/InAs(100) film along the four major crystal-axes. The films clearly display a cubic anisotropy, with the magnetic easy axes along $\langle 001 \rangle$, the easy axes of bulk bcc Fe. The coercivity, H_c, of the 50 ML film is 28 ± 1 Oe, which is very similar to that of Fe/GaAs; e.g., $H_c = 25$ Oe and $H_c \sim 30$ Oe were reported by Florczak and Dahlberg [4], and Gester et al. [5], respectively.

The magnetization of an Au(20 ML)/Fe(100 ML)/InAs film has been measured using AGFM. The magnetic moment $1.7 \pm 0.2 \times 10^3$ emu/cm^3 of the film is comparable

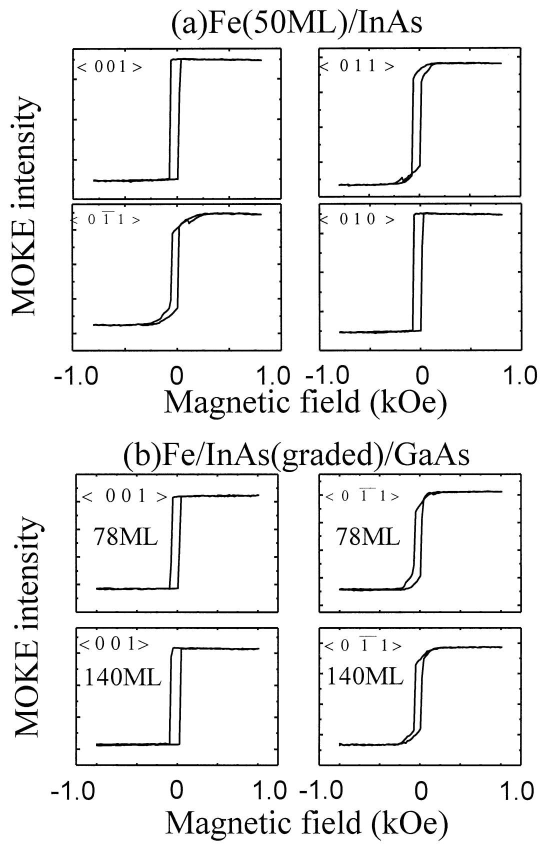

Fig. 4. In situ MOKE hysteresis loops of (a) Fe(50 ML)/InAs(100), and (b) Fe(78 ML, and 140 ML)/InAs(graded)/GaAs(100) for the magnetic field applied along four or two major crystal-axes.

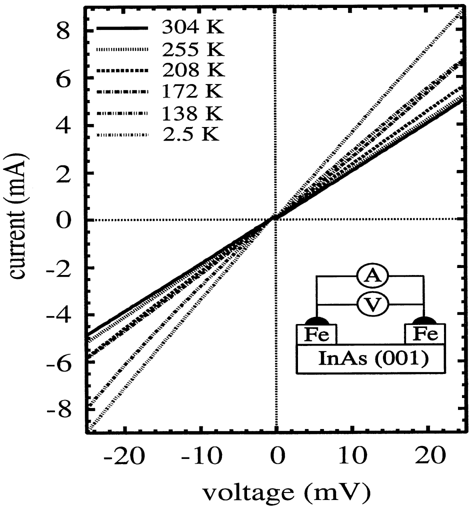

Fig. 5. $I-V$ measurements of Fe(50 ML)/InAs(100) in the temperature range 2.5 to 304 K. The dimension of the Fe contacts, processed using optical lithography, is 50 μm with separation of about 250 μm.

with that of bulk Fe. It is interesting to note that the magnetizations of Fe/GaAs and Fe/InAs when both are grown at 175°C are very different. The magnetization is only 1.0×10^3 emu/cm^3 emu, reduced by 40%, for a Fe film of 96 Å (67 ML) on GaAs [3]. This indicates that Fe moments near the interface in Fe/InAs are much larger than that in Fe/GaAs under the same growth condition.

Fe is expected to form an ohmic contact to InAs due to the pinning of the Fermi energy in the conduction band at the InAs surface, which results in a charge accumulation layer at the surface. In order to verify this, the samples were characterized by current–voltage ($I–V$) measurements. The Fe contacts, with dimensions of approximately 50 μm, were processed using optical lithography and a combination of CHF_3 based reactive ion etching and selective wet etching. The substrate is n-type InAs with $\sim 2.5 \times 10^{18}$ cm^{-3} sulfur doping. Typical $I–V$ characteristics are shown in Fig. 5. They are linear over the temperature range 304 to 2.5 K and show no indication of a Schottky barrier. The equivalent resistance is weakly dependent on temperature, varying from 5 to 2.8 Ω in the temperature range 304 to 2.5 K. This is due to the increase of the mobility at low temperature. In conclusion, we have successfully grown single crystal Fe films which form low-resistance ohmic contacts on InAs(100).

5. Fe/In_xGa_{1-x}As(50 nm)/GaAs(100)

To minimize the effect of the Schottky barrier in GaAs based devices, a graded layer of n-In_xGa_{1-x}As from GaAs to InAs was grown on GaAs before FM deposition. The ferromagnetic metal should then form an ohmic contact without the need to diffuse the contacts into the active region of the devices, which would create an amorphous layer. Although GaAs and InAs are not lattice matched (7% mismatch), a smooth alloy transition over a thickness of 50 nm was found to produce a relatively uniform layer [14]. The substrate was capped with As and then transferred to the UHV chamber for the deposition. The As capping layer was desorbed before Fe growth. The temperature was kept at 175°C during Fe growth. The magnetic and structural properties have been studied in detail during growth [15]. LEED measurements showed that Fe grew epitaxially on this graded InAs buffer layer. In situ MOKE in Fig. 4b showed a distinct cubic anisotropy with the easy axis along ⟨001⟩, which confirms that well ordered single crystal Fe films have been stabilised on this InAs (graded)/GaAs substrate.

6. Spin-polarized FET

Dattas and Das [2] have suggested the construction of a spin-polarized field effect transistor (FET), which applied the spin-injection concept to a semiconductor. Such devices have not yet been demonstrated as far as we know. This may be due to the difficulty in achieving a low resistance contact between the FM pad and the two dimensional electron gas (2DEG). Based on the growth studies given above, a spin FET device is being developed in our group. Fig. 6 shows a schematic diagram of this device. The spin-polarized carriers are injected and collected by single crystal Fe elements, in which the electrons have a long coherence length. With an InAs graded buffer layer, there is no energy barrier for the electron to tunnel through FM–2DEG–FM channel. The device in Fig. 6 can operate

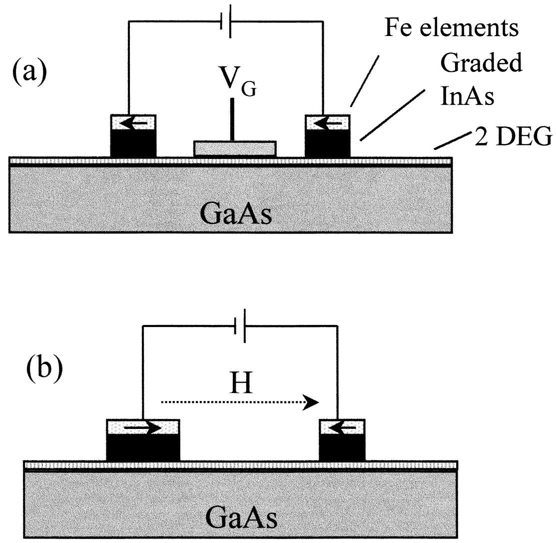

Fig. 6. Schematic diagram of the spin-polarized field effect transistor (FET) using graded InAs buffer layer and single crystal Fe pads, (a) control by external electric field, and (b) control by the external magnetic field.

in two modes; (a) Control by external electric field—a gate voltage is applied to the 2DEG to alter the spin precession, as proposed by Prinz [16]. This will allow control of the alignment of the electron's spin with respect to the magnetization vector of the second FM pad so as to modulate the current. (b) Control by external magnetic field—switching of each Fe pad could be controlled separately by applying an external magnetic field if the two Fe pads have different shapes. It has been previously shown that the coercivity of the ferromagnetic wires depends on the square ratio due to the influence of shape anisotropy [17,18]. The resistances of the device could be different for two different configurations, namely parallel and anti-parallel alignment of the magnetization orientations if the separation of the two FM pads is smaller than the spin-diffusion length in the 2DEG. The fabrication of such elements in both single crystal Fe and FeNi has been well demonstrated in our laboratory [17,18]. The spin-diffusion length of electron in 2DEG is still an open issue of current interest.

7. Conclusion

We demonstrate that with careful substrate preparation and suitable growth conditions Fe/GaAs(100) films do not exhibit a magnetically 'dead' layer at the interface. We have also shown that the metal/semiconductor heterostructure Fe/InAs(100) may have suitable electrical and magnetic properties for the fabrication of future magnetoelectronic devices, and that the favourable properties of this interface may be taken advantage of in GaAs based devices by using an InAs(graded)/GaAs(100) substrate.

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