EUROPHYSICS LETTERS 1 March 2000

Europhys. Lett., **49** (5), pp. 651–657 (2000)

Nano-patterning of magnetic anisotropy by controlled strain relief

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(received 20 July 1999; accepted in final form 10 December 1999)

PACS. 75.30.Gw – Magnetic anisotropy.
PACS. 61.72.Ff – Direct observation of dislocations and other defects (etch pits, decoration, electron microscopy, X-ray topography, etc.).
PACS. 75.70.Ak – Magnetic properties of monolayers and thin films.

Abstract. – In the highly strained system Fe/W(001) the formation of parallel dislocation bundles upon nucleation of fifth layer islands is used to locally break the fourfold symmetry. The uniaxial strain relief in the dislocation bundles introduces strong uniaxial in-plane magnetic anisotropies. By controlling the density of fifth layer islands local magnetic anisotropies are structured on the nanometer scale. As a result of this patterning of anisotropies, the magnetic properties of the films are drastically altered. As a function of the pattern size, the coercivity of the films can be varied in a controlled way over more than two orders of magnitude without changing the film thickness. For pattern sizes larger than the estimated domain wall width, MOKE and micromagnetic calculations indicate the break-up of the film into magnetic in-plane structures on the 100 nm scale.

During the past decade many studies have been devoted to the magnetism of structures of reduced dimensions. Especially ultrathin films exhibit a richness of behaviour often related to surface or interface anisotropies [1]. By choosing specific film and substrate combinations, these anisotropies may be tailored such that contrary to shape anisotropy a perpendicular magnetisation is favoured [2]. Also in-plane anisotropies may be varied by selecting substrates of different symmetries [3–5], using vicinal surfaces as substrates [6–8] or by adsorption [9,10]. Except for few special cases [11–13], these methods of influencing magnetic anisotropy are operating on the thin film as a whole and are of limited use in the view of lateral structuring of magnetic properties.

In this letter we present a new approach that allows the structuring of in-plane magnetic anisotropy on nanometer scale by partial strain relief in ultrathin magnetic films. The large tensile strain of 10.4% in bcc Fe films on W(001) is shown to be uniaxially relieved by the formation of dislocation bundles along $\langle 100 \rangle$ directions resulting in a local breakdown of the fourfold symmetry and introduction of uniaxial magnetic anisotropies. Under suitable conditions, growth leads to the self-organisation of a network of locally varying anisotropies allowing the controlled structuring of magnetic anisotropies on the nanometer scale and by this to tailor technically relevant macroscopic as well as microscopic properties of the magnetic films.

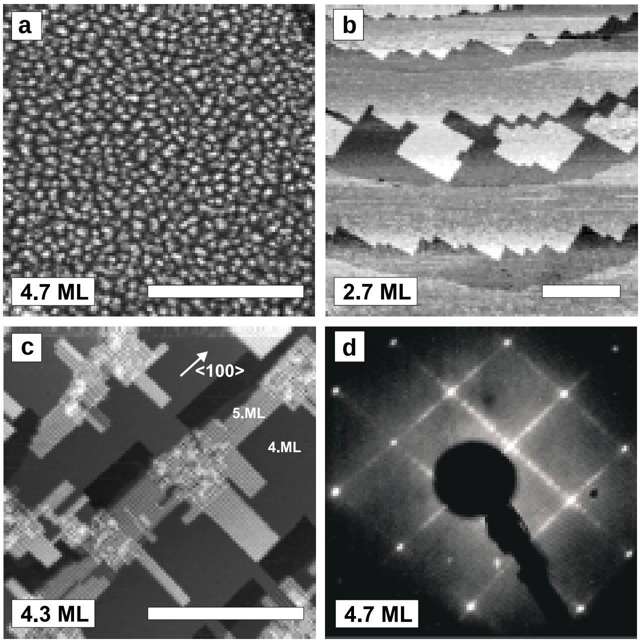

Fig. 1 – Morphology (a-c) and LEED diffraction pattern (d) of Fe films on W(001) of thicknesses as indicated. Deposition was carried out at 300 K (a) and $\approx 400\,\mathrm{K}$ (b-d) with a rate of $\approx 2\,\mathrm{ML/min}$. The white bars represent 100 nm (a-c). Electron energy: 166 eV (d).

Experiments were carried out in ultrahigh vacuum (UHV) at a base pressure of 5×10^{-11} mbar. The W(001) sample was cleaned by cycles of glowing in O_2 ($\approx 1700\,\mathrm{K}$, 10^{-7} mbar) and flashing to $\approx 2500\,\mathrm{K}$ in the absence of O_2 until no contaminations were detected by Auger electron spectroscopy and low energy electron diffraction (LEED) showed sharp (1×1) diffraction patterns. Scanning tunneling microscopy (STM) images revealed clean, flat terraces of sizes larger than 100 nm separated by single atomic steps. Fe (99.999% purity) was deposited by electron beam evaporation. Coverages were calibrated in pseudomorphic monolayers (ML) using STM and were crosschecked by medium energy electron diffraction from Fe/W(110). During growth, the sample was heated to different temperatures measured with a thermocouple attached to the sample holder in close vicinity of the sample. After growth, *in situ* characterisation of the film structure, morphology and magnetism was carried out by LEED, STM and longitudinal magneto optic Kerr effect (MOKE), respectively, at room temperature.

First, we concentrate on the growth of and dislocation formation in ultrathin Fe films on W(001), especially at intermediate growth conditions between the two limiting cases of low and high atom mobility. Fe grows in its bcc modification on W(001) with parallel orientation of the $\langle 100 \rangle$ directions of Fe and W. At room temperature, a low mobility leads to the growth of rough films consisting of small ($\approx 10\,\mathrm{nm}$), three-dimensional islands (see fig. 1a). LEED images show a blurred (1×1) structure in accord with rough growth. Strain relaxation by introduction of misfit dislocations gradually sets in around 3 ML total coverage as stress measurements revealed [14, 15]. At temperatures above $\approx 600\,\mathrm{K}$, mobility is high enough to allow thermodynamic equilibration of the film. We find Stranski-Krastanov growth —fully relaxed three-dimensional crystallites on top of a 2 ML pseudomorphic Fe carpet [18]— in

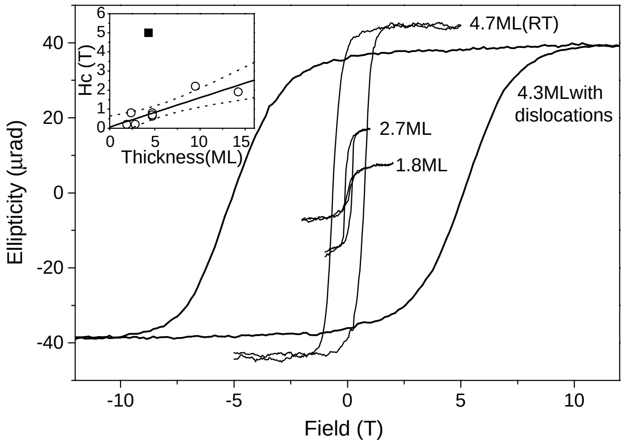

Fig. 2 – MOKE magnetisation loops of Fe films of thicknesses as indicated taken along ⟨110⟩ direction at 170 K (1.8 ML) and 300 K (> 2.7 ML). For more details see text. The inset shows typical coercivities of pseudomorphic and room temperature grown films (open circles) in comparison to a film with dislocation bundles (filled box) and a linear fit with a 1σ confidence region (dotted lines).

agreement with previous work [16,17]. At intermediate temperatures around 400 K, however, STM reveals layer-by-layer growth. The mobility of the atoms is high enough to form large, flat islands or to attach to a descending step edge but not high enough to form three-dimensional crystallites (see fig. 1b). Interestingly, sharp (1×1) LEED patterns are observed up to 4 ML coverage indicating the growth of pseudomorphic, fully strained Fe films. Obviously, strain relaxation in this system is much delayed in comparison to Fe/W(110), where dislocation formation already sets in around 1.5 ML [19,20]. Qualitatively, this can be understood when realising that in Fe/W(110) the film plane is a glide plane of bulk bcc Fe [21] and thus in-plane strain can be easily released while for Fe/W(001) the film plane is not a glide plane of bulk Fe [21]. In the latter case the strain can only be released by gliding under an angle with respect to the film plane which is kinetically hindered. However, when the coverage exceeds 4 ML, the kinetic barrier is overcome and dislocation lines are formed along ⟨100⟩ directions that show up as white lines (topographical protrusions) in the STM images (fig. 1c) —similar to those of the Au(111) reconstruction [22]. Strain relief is incomplete and exclusively takes place underneath fifth layer islands, *i.e.* patches, where the film is locally 5 ML thick. A large fraction of these islands displays only uniaxial strain relief by bundles of parallel dislocations. This is further confirmed by the occurrence of a two-domain (9×1) LEED pattern (see fig. 1d). The parallel dislocation bundles along the two equivalent ⟨100⟩ directions are separated by exactly 9 atomic distances on the surface. Interestingly, the fifth layer islands are elongated along the dislocation lines, *i.e.* the fully strained direction, and not perpendicular, *i.e.* the relaxed direction, as should be favoured by the elastic energy. Obviously, the islands are not in their equilibrium shape but their shape is dictated by growth kinetics. It can be concluded, that it is harder to nucleate a dislocation than to prolong an existing one. By this, dislocations in fifth layer islands travel like zippers through the film when the tensile strain is relieved and additional Fe atoms are incorporated into the film at the end of the dislocations.

Now, we correlate the film structure with its macroscopic magnetic properties as measured with MOKE. In accordance with some experiments [17,23,24] and calculations [25] we find no magnetisation in monolayer films at 150 K and observe the onset of magnetisation at around