Origin of transverse magnetization in epitaxial Cu/Ni/Cu nanowire arrays

M. Ciria,1,2 J. F. Castaño,3 J. L. Diez-Ferrer,2,4 J. I. Arnaudas,4,5 B. G. Ng,3 R. C. O’Handley,3 and C. A. Ross3
1Departamento de Magnetismo de Sólidos, Instituto de Ciencia de Materiales de Aragón, Consejo Superior de Investigaciones Científicas, 50009 Zaragoza, Spain
2Departamento de Física de la Materia Condensada, Universidad de Zaragoza, 50009 Zaragoza, Spain
3Department of Materials Science and Engineering, Massachusetts Institute of Technology, Cambridge, Massachusetts 02139, USA
4Instituto de Nanociencia de Aragón, Universidad de Zaragoza, 50009 Zaragoza, Spain

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The patterning-induced changes in the magnetic anisotropy and hysteresis of epitaxial (100)-oriented Cu/ Ni(9, 10, 15 nm)/Cu planar nanowires have been quantified. When the Ni films are patterned into lines, strain relaxation leads to a thickness-dependent net in-plane anisotropy transverse to the lines. The magnetoelastic anisotropy was found from the three-dimensional strain state measured directly by synchrotron x-ray diffraction and has a value of −21 kJ/m² for 10-nm-thick nanowires. The angular dependence of the remanence of the nanowires indicates that the in-plane easy direction is the result of the competition between the cubic magnetocrystalline anisotropy and a uniaxial anisotropy that includes shape and magnetoelastic effects. The patterning-induced changes in magnetoelastic anisotropy, combined with the shape and magnetocrystalline anisotropies, quantitatively explain the net anisotropy of the nanowires. Thus by controlling the film thickness and wire orientation, the easy axis direction of the nanowires may be controlled.

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I. INTRODUCTION

The control of the magnetic anisotropy and domain configurations in lithographically defined magnetic elements with submicron lateral dimensions is of importance for spintronic devices. Planar nanowires (NWs), in particular, have been investigated because of their potential use in domain-wall devices proposed for data storage1 and logic applications.2 Most reported work has been focused on nanowires patterned from polycrystalline films, whose properties are dominated by shape anisotropy, and which exhibit domains with the magnetization pointing along the wire axis separated by either transverse or vortexlike 180° head-to-head domain walls (DWs).3 However, materials with different domain states and DW geometries could offer advantages in spintronic applications, for example, metallic planar NWs with perpendicular magnetic anisotropy exhibit current-induced DW motion at lower threshold current values compared to those of NWs with in-plane anisotropy.4,5 Therefore, developing methods for tuning the magnetic anisotropy of NWs may offer new opportunities for spintronic devices.

The magnitude and direction of the net anisotropy determine the domain structure and the DW type of a patterned element. Anisotropy is most commonly controlled via the geometrical shape of the structure,6 or by magnetocrystalline anisotropy,7 but in certain strained epitaxial magnetic films, magnetoelastic (ME) effects can dominate the anisotropy,8–10 leading, for example, to a net anisotropy $K_p$ oriented perpendicular to the film plane. Patterning of a strained layer breaks the in-plane symmetry and produces an asymmetric strain relaxation and an in-plane ME anisotropy contribution to the total magnetic anisotropy energy.

The in-plane anisotropy energy of a NW made from a single-crystal cubic material includes the magnetocrystalline anisotropy $K_1 \cos^2 \phi \sin^2 \phi$ plus the ME and magnetostatic contributions $(K_{me} + K_{sh})\sin^2 \phi$, a uniaxial term, where $\phi$ is the angle between the magnetization $M$ and the NW axis. The shape energy term can be written as $K_{sh} = 0.5 \mu_0 (N_x - N_y) M^2$ (we assign $x$ NW axis and $y \perp$ NW axis, thus $N_x$ and $N_y$ are, respectively, the transverse and longitudinal demagnetization factors). Because $K_{sh}$ is always positive (giving an easy direction along the NW axis), the in-plane uniaxial anisotropy term will depend on the sign of the ME coefficient $K_{me}$. If $K_{me} > 0$ the uniaxial anisotropy along the NW is enhanced, as observed in magnetic semiconductors such as (Ga,Mn)As/(In,Ga)As,11 while if $K_{sh} < -K_{me}$ the uniaxial anisotropy drives $M$ transverse to the NW axis. For cubic thin-film materials grown on the (001) plane, $K_{me} = -B_1 \epsilon_x$, where $B_1$ is the ME stress coefficient and $\epsilon_x = \epsilon_{xx} - \epsilon_{yy}$ is the strain associated with the breaking of the in-plane fourfold symmetry. The sign of $\epsilon_x$ in a patterned heteroepitaxial system can be obtained by considering two factors: first, the strain relaxation in a NW is larger along the direction transverse to the NW axis so $|\epsilon_{xx}| > |\epsilon_{yy}|$ and second, the misfit between the substrate lattice parameter $a_s$ and the lattice parameter of the magnetic material $a_m$ determines the sign of $\epsilon_{xx}$ and $\epsilon_{yy}$, so that $a_m > (>) a_s \Rightarrow \epsilon_{xx}, \epsilon_{yy} < (>) 0$ and $\epsilon_x < (>) 0$. The Ni/Cu(001) system, which has $B_1 > 0$ and $a_{Cu} < a_{Ni}$ can meet the criterion $-K_{me} = B_1 \epsilon_x > K_{sh}$ and is therefore able to produce NWs with magnetization in-plane but transverse to the wire axis. We have shown previously that patterning Cu/Ni/Cu(001) epitaxial films of 6.9 and 20.6 nm Ni thickness can indeed lead to NWs with a net transverse anisotropy.12

In the present work we demonstrate, by measuring the in-plane lattice parameters and determining $\epsilon_x$, that the transverse anisotropy observed in NW arrays of patterned epitaxial Cu/Ni/Cu films with nickel thickness $t_{Ni} = 9, 10$, and 15 nm, is due to a $K_{sh}$ that is large enough to overcome $K_{me}$. We show that the easiest in-plane direction forms an angle with the NW axis that depends on the ratio $K_t / K_u$ with $K_t$ the cubic magnetocrystalline anisotropy constant and $K_u = (K_{me} + K_{sh})\sin^2 \phi$, the uniaxial term including ME and

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II. EXPERIMENTAL METHODS

Epitaxial Cu (5 nm)/Ni (t_{Ni})/Cu (100 nm) films with t_{Ni} between 2 and 15 nm were grown on Si (001) wafers at room temperature by electron-beam evaporation in a chamber with a base pressure below 2 \times 10^{-10} Torr, using a procedure reported elsewhere.\(^{13}\) The single-crystal nature of each layer and the epitaxial relationships, i.e., Si(100) ||Cu(110)|| Ni(110), are observed in situ by reflection high-energy electron diffraction (RHEED), see Figs. 1(a) and 1(b), and high-resolution grazing incidence x-ray diffraction (XRD), done in the BM25B beamline of ESRF using a photon energy of h\nu=15 keV. The magnetic properties of the samples were investigated by alternating gradient magnetometry, vibrating sample magnetometry, and magnetic force microscopy (MFM), using a low-moment tip. In addition to the anisotropy calculation based on strain, an estimate of the effective magnetic anisotropy constants was also made from the difference between magnetization loops measured along specific directions.\(^{10}\) To eliminate the effect of hysteresis, the anhysteretic M-H curve is calculated by averaging the descending and ascending branches of the hysteresis loop at constant values of M. Arrays of Cu/Ni/Cu nanowires with the wire axis along the Ni (100) direction were fabricated by ion-beam etching (see Fig. 2), using a Ta grating as a hard

shape anisotropy. \(K_1\) favors magnetization along the in-plane (110) directions and \(K_2\) favors anisotropy transverse to the NW axis. Thus, the combination of competing anisotropies suggests a way to engineer NWs with controlled easy magnetization direction.

![Image](image-url)
and a wire width of 600 nm. and 200 nm, respectively, while 4NW had a period of 800 nm and a wire width of 300, 200 and 150 nm, which will be labeled as 4NW, 9NW, 10NW, and 15 NW, respectively. Samples 9NW, 10NW, and 15NW have a period close to 450 nm and a wire width of 300, 200 and 200 nm, respectively, while 4NW had a period of 800 nm and a wire width of 600 nm.

III. RESULTS AND DISCUSSION

Magnetometry shows that the unpatterned nickel films exhibit out-of-plane magnetization, with an effective perpendicular magnetic anisotropy $K_p>0$, for $t_{Ni}$ between about 1.5 and 11.5 nm [see Fig. 1(c)], in agreement with other work. The 4-, 9-, and 10-nm-thick films are more easily magnetized out-of-plane compared to in-plane ($K_p>0$) while for the 15 nm film the opposite behavior is observed ($K_p<0$).

Figure 2 shows data for the patterned films. The in-plane loops with $H$ applied along the $x$ and $y$ directions provide information about the effects of patterning on the magnetic behavior. These samples span the range from high $K_p$ ($t_{Ni}=4$ nm) to a regime ($t_{Ni}=9$, 10, and 15 nm) where $K_p$ is low and its sign changes from positive to negative. For 4NW, $K_p$ is positive, while for 9NW, 10NW, and 15NW, $K_p$ is negative, showing that $K_p$ has changed sign for $t_{Ni}=9$ and 10 nm. For the three samples with negative $K_p$ the hysteresis loops suggest that $M$ is oriented preferentially in plane, transverse to the nanowires, particularly for the 10NW and 15NW samples. This result, which is consistent with our prior work, is contrary to what would be expected from shape anisotropy alone, and demonstrates the governing effect of $K_{me}$.

A direct measurement of the strain asymmetry is required to quantify the ME anisotropy. This was performed for sample 10NW. The strain state in the 10NW array was determined from XRD reciprocal space mapping in the vicinity of the asymmetric (402) and (042) Bragg reflection from Cu and Ni layers (see Fig. 3). These scans are sensitive to the lattice parameter $a_x$ parallel to the NW axis [see Fig. 3(a)] and $a_y$ transverse to the NW axis [see Fig. 3(b)], as well as to the out-of-plane lattice parameter. These scans show both Ni and Cu reflections, and while the copper peak is at the bulk value (units in Fig. 3 are related to the bulk copper lattice parameter) the nickel peak positions are both shifted with respect to the bulk value and by about 0.15 units with respect to each other. From the experimental positions of the maxima of the scattered intensity the in-plane lattice parameters were determined as $a_x=3.553$ Å, $a_y=3.541$ Å, and $a_z=3.504$ Å leading to $\epsilon_{xx}=(8.3\pm0.1)\times10^{-3}$, $\epsilon_{yy}=(4.9\pm0.1)\times10^{-3}$, and $\epsilon_{zz}=(-5.7\pm0.1)\times10^{-3}$, respectively. Figure 4 gives a comparison between these in-plane strain components and those obtained for a series of unpatterned films measured in the same diffractometer by grazing incidence XRD, as well as data reported elsewhere.

The $\epsilon_{xx}$ of the 10NW sample is close to the value found for the unpatterned films while $\epsilon_{yy}$ is about half as large, indicating that the nanowires undergo a relaxation transverse to the NW axis. Considering the $z$ direction, $\epsilon_{zz}$ is smaller than the value $\epsilon_{zz}=-12.8\times10^{-3}$ calculated for the biaxially stressed unpatterned film ($\epsilon_{xx}=2(c_{12}/c_{11})\epsilon_{xx}$, where the $c$s are the nickel elastic constants and $\epsilon_{xx}=-10\times10^{-3}$), showing that $\epsilon_{zz}$ is also relaxed in the NW array.

Based on $\epsilon_z=\epsilon_{xx}-\epsilon_{yy}=3.4\pm0.2\times10^{-3}$ for sample 10NW, we obtain $K_{me}=-21\pm2$ kJ/m$^2$ [with $B_I=6.2$ MPa (Ref. 16)]. $K_{sh}$ was determined by using the demagnetizing factors for arrays of planar NWs reported in Ref. 17. With $N_{y}=0$ and $N_{z}=0.91t_{Ni}/w$ ($w=$ width) for the array, a value slightly smaller than that of a single infinite NW, $K_{sh}=6.6$ kJ/m$^3$ for sample 10NW. These values imply $K=\epsilon_{xx}-14.4$ kJ/m$^3$ which quantifies the transverse anisotropy
in terms of a ME effect due to asymmetrical relaxation of the in-plane epitaxial lattice strain.

An estimate of \( K_u \) may also be obtained from the anhysteretic \( M-H \) curves measured in the longitudinal and transverse directions.\(^{10} \) This yields \( K_u \) (15NW) = \(-6.0 \) kJ/m\(^3\); \( K_u \) (10NW) = \(-6.3 \) kJ/m\(^3\); and \( K_u \) (9NW) = \(-4.0 \) kJ/m\(^3\).

The value for 10NW is in reasonable agreement with that calculated from the measurement of the strain, considering the approximations of the estimate based on hysteresis. The NWs were patterned such that the (100) in-plane directions are parallel to the NW axis, and the (110) in-plane directions, which are magnetically easy according to the magnetcocrystalline anisotropy, form an angle of 45° to the NW axis. Thus, a competition between the uniaxial term and the magnetocrystalline anisotropy is expected. To elucidate that competition we examine the symmetry of the in-plane magnetcocrystalline contribution is expected. The arrows indicate particles used as reference. The continuous lines indicate the spaces between Ni NWs.

For the 9NW and 10NW arrays the in-plane (110) direction is no longer the easy in-plane direction. Instead, the easy and (100) hard axes, respectively, defined by the magnetocrystalline anisotropy \( K_1 \cos^2 \phi \sin^2 \phi \) with \( K_1 = -4.5 \) kJ/m\(^3\).\(^{16} \)

The minimum \( K_u \) for the 9NW and 10NW arrays the in-plane \( \phi = 0° \) corresponds to a (100) direction. As expected, for the unpatterned film, \( M_r/M_s \) displays a fourfold dependence on \( \phi \). The (110) directions, at 45° to the NW axis, are the in-plane easy axes. \( M_r/M_s \) takes maximum and minimum values along the (110) easy and (100) hard axes, respectively, defined by the magnetocrystalline anisotropy \( K_1 \cos^2 \phi \sin^2 \phi \) with \( K_1 = -4.5 \) kJ/m\(^3\).\(^{16} \)

Taking the value at which \( M_r/M_s \) has its maximum value as the minimum of \( E_{\text{ani}}(\phi) \) and using the value of \( K_u \) obtained from hysteresis loops and strain data, we obtain, \( K_1 \sim 7.5 \) and \(-18.4 \) kJ/m\(^3\), respectively, for sample 10NW. This suggests that \( K_1 \) is greater than the magnetocrystalline anisotropy of bulk Ni, which is \(-4.5 \) kJ/m\(^3\). An enhancement of \( K_1 \) in nickel thin films has been reported\(^{18,19} \) and its origin linked to a magnetoelastic contribution due to the isotropic in-plane stress in the film arising from the misfit between substrate and film. The lines in Fig. 2(g) represent \( E_{\text{ani}}(\phi) \) calculated for the 9NW and 15NW arrays, and show excellent agreement with the measured angular dependence of the ratio \( M_r/M_s \) indicating the presence of a uniaxial term in \( E_{\text{ani}} \).

Magnetic force microscopy was used to clarify the domain structure and aid in the interpretation of the anisotropy data. Figure 5 shows images of samples 4NW and 9NW. For 4NW, comparing images taken in the same area, in the remanent state [Fig. 5(a)] after out-of-plane saturation the Ni NWs show all-bright magnetic contrast, indicating a high remanence consistent with the hysteresis loop, while the out-of-plane ac-demagnetized state [Fig. 5(b)] shows dark and bright domains along the wires with lengths of 1–10 μm, much longer than the wire width or thickness. For sample 9NW, Fig. 5(c), taken also in a out-of-plane ac-demagnetized state, the domain configuration is more complicated and a few μm long domains are visible with dark and light contrasts. Using dust particles that induce a topography signal for reference, see Fig. 5(d), we observe that the maximum magnetic contrast corresponds to near the edges of the topography, and is consistent with a magnetization transverse to the wire length, as suggested by the hysteresis loops. A simple model supports the qualitative analysis of the MFM images: the strength of the \( z \) component of the fringing field as function of \( \gamma \) for an infinite planar nanowire with uniform magnetization confined in the \( y-z \) plane transverse to the wire axis is calculated\(^{16} \) as a function of the angle \( \gamma \) between

![FIG. 5.](image)

![FIG. 6.](image)
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IV. SUMMARY

We have quantitatively analyzed the magnetoelastic, shape and magnetocrystalline anisotropies for planar nanowire arrays made from heteroepitaxial Cu/Ni/Cu thin films with a range of Ni film thicknesses. Patterning the continuous films leads to anisotropic stress relief and a magnetoelastic anisotropy which can dominate the net magnetic anisotropy of the Ni nanostructures. These results show how magnetoelastic effects may be used to control the net anisotropy and magnetic properties of patterned nanostructures, leading, for example, to nanomagnets with tailored in-plane magnetic anisotropy.

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M and the z direction, Fig. 6. For γ=0, the stray field is maximum in the nanowire center, as is observed for 4NW, while if γ increases, the extremal values move toward the NW edge as seen for 9NW.

Finally, we reconsider the strain data to show why \( K_p \) decreases on patterning: \( K_p \) includes the ME term \(-B_1(\varepsilon_{zz}-0.5(\varepsilon_{xx}+\varepsilon_{yy}))\) and the components of strain decrease on patterning. For an unpatterned Ni film in biaxial stress, the ME contribution to anisotropy reduces to \( B_1\varepsilon_{zz}[1+2(c_{12}/c_{11})] \), with \( B_1=6.2 \text{ MPa} \) and the ratio \( c_{12}/c_{11}=0.64 \). This gives a value for the ME contribution of 141 kJ/m\(^3\) for the 10-nm-thick film in which \( \varepsilon_{zz}=0.01 \). For the 10NW sample, we use for the strain components the values measured by x-ray diffraction. We find that the ME contribution to \( K_p \) is 76 kJ/m\(^3\), showing that \( K_p \) must have decreased by about 65 kJ/m\(^3\) upon patterning. This decrease is larger than the net \( K_p \) of the unpatterned film, which was about 10 kJ/m\(^3\). This explains why \( K_p \) becomes negative in the 9 and 10-nm-thick patterned NWs despite being positive for the unpatterned films. In contrast, for the 4-nm-thick sample, the \( K_p \) of the unpatterned film was about 90 kJ/m\(^3\) so the strain relaxation is insufficient to drive the net anisotropy in plane, and the patterned 4NW sample retains its out-of-plane anisotropy.

*e@ciria@unizar.es


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