Magnetic Surface Anisotropy of Transition Metal Ultrathin Films

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Abstract. In order to study the magnetic anisotropy of transition metal ultrathin films, we have performed tight-binding calculations including spin-orbit coupling. Besides the anisotropy energy these calculations also yield the orbital moment, which turns out to be much more anisotropic than in bulk materials. The effects of interfacial mismatch and roughness are discussed within phenomenological models. We also briefly review experimental results on the magnetic surface anisotropy (MSA) in transition metal ultrathin films. In some cases such as Au/Co/Au(111) sandwiches the MSA wins the competition with the shape anisotropy arising from the magnetostatic energy: below a critical thickness this leads to a perpendicular spontaneous magnetization. We show the effects of this crossover on the hysteresis loops and on the magnetoresistance, and the effects of interface roughness on the critical thickness.

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The very thin ferromagnetic films are of great current interest because their magnetic properties differ from those of the bulk ferromagnets. For instance, they can show enhanced or reduced atomic magnetic moments and thickness dependent Curie temperature. One of the most attractive new features, both for fundamental physics and for potential applications, is the magnetic anisotropy induced by the symmetry breaking at the surfaces of the film.

In a ferromagnetic ultrathin film, the anisotropy energy (per unit volume) can very often be written [1–3] as

\[ E_a = K_{sat} \sin^2 \theta, \]

where \( \theta \) is the angle between the normal to the plane and the magnetization. Moreover

\[ K_{sat} = K_{sat}^{eff} + \frac{2K_{sat}^{eff}}{t}, \]

where \( t \) is the thickness, \( K_{sat}^{eff} \) is the effective volume anisotropy constant (containing the demagnetizing term as well as the magneto-crystalline terms) and \( K_{sat}^{eff} \) is an effective surface or interface anisotropy constant. Our aim is to discuss the physical origin of this surface anisotropy.

As suggested by Néel [4], the lack of neighbours at a surface can give rise to a so-called magnetocrystalline surface anisotropy. This anisotropy has been studied by means of first principles calculations by other authors [5–7]. In Sect. 1 we present tight-binding calculations, including spin-orbit coupling, for monolayers [8]. These calculations yield not only the magnetocrystalline surface anisotropy but also the orbital magnetic moment.

Section 2 treats the effect of roughness, both on the dipolar shape anisotropy and on the magnetocrystalline surface anisotropy [9–10].

In Sect. 3 we discuss the effect of the strains due to lattice mismatch between the substrate and the film and show that they can induce a substantial magnetoelastic surface anisotropy [3].

Finally in Sect 4 we review some experimental results on surface anisotropy in ferromagnetic films of transition metals. In particular we discuss the case of hcp Co on Au(111), for which the surface anisotropy
favors an easy axis perpendicular to the film plane and is strong enough to overcome the demagnetizing energy below a critical thickness.

1. Magnetocrystalline Anisotropy and Orbital Moment of Transition-Metal Monolayers

The method we have used is a perturbative tight-binding one; thus it cannot achieve the same accuracy as ab initio calculations [5-7]. It is however very useful for providing correct orders of magnitude. The main steps of the method are summarized here (for the details see [8]).

Magnetocrystalline anisotropy and unquenching of the orbital moment are due to the spin-orbit coupling, given by:

\[ H_{SO} = \zeta \mathbf{L} \cdot \mathbf{S}. \]

We first calculate the (spin-polarized) ground state electronic structure within the tight-binding scheme and using the Hartree-Fock approximation. The spin orbit coupling is then introduced as a perturbation.

The corrections to the energy and ground-state wavefunction respectively given by the well-known formulae:

\[ \delta E = \sum_{\text{exc}} \left( \frac{|\langle \text{gr} | H_{SO} | \text{exc} \rangle|^2}{E_{\text{gr}} - E_{\text{exc}}} \right), \]

\[ \delta |\text{gr}\rangle = \sum_{\text{exc}} \left( |\langle \text{exc} | H_{SO} | \text{gr} \rangle| \frac{E_{\text{gr}} - E_{\text{exc}}}{E_{\text{gr}} - E_{\text{exc}}} \right), \]

where \(|\text{gr}\rangle, |\text{exc}\rangle, E_{\text{gr}}, E_{\text{exc}}\) are respectively the ground state, excited state and corresponding energies for the unperturbed system. The average value of the \(L\) component parallel to the spin quantization direction is:

\[ \langle L_\perp \rangle = \sum_{\text{exc}} \left( |\langle \text{exc} | H_{SO} | \text{gr} \rangle| \frac{E_{\text{gr}} - E_{\text{exc}}}{E_{\text{gr}} - E_{\text{exc}}} \right) + \text{c.c.} \]

The matrix elements \( |\text{exc} | H_{SO} | \text{gr} \rangle \) depend on the orientation of the spin-quantization axis, i.e. on the magnetization direction: this is the physical origin of the magnetocrystalline anisotropy.

Equations (4, 6) clearly indicate the strong connection between magnetocrystalline anisotropy and orbital moment. More precisely, one obtains [8]

\[ \delta E = K_0 + K_2 \sin^2 \theta, \]

\[ \langle L_\perp \rangle = L_0 + L_2 \sin^2 \theta, \]

where \(K_2\) is the magnetocrystalline anisotropy energy, \(L_0\) and \(L_2\) are the isotropic and anisotropic parts of the orbital moment respectively; \(K_2\) and \(L_2\) are related by:

\[ K_2 \approx \frac{\zeta}{4} L_2. \]

Accordingly, with an anisotropy energy of \(10^{-3}\) eV/atom as observed in ultrathin films (Sect. 4) and \(\zeta = 0.05\) eV, one can obtain a magnetic moment anisotropy of the order of 0.1 \(\mu_B/\text{atom}\).

This magnetic moment anisotropy is much larger than in bulk materials. To our knowledge, it was never predicted before. Actually, (9) holds for bulk materials, too, since for Ni and Fe, one has \(\Delta E \approx 10^{-6}\) to \(10^{-3}\) eV/atom, \(\Delta M \approx 10^{-4} \mu_B/\text{atom}\) so that \(\Delta E/\Delta M \approx \zeta\).

Table 1 displays the anisotropy energy \(K_2\), the in-plane and out-of-plane orbital magnetic moments \(M_I\) and \(M_P\) calculated for fcc(001) and fcc(111) monolayers of transition metals for various numbers of valence electrons \(N_e\).

Because the present model of a perfectly flat, unsupported monolayer is far from any real system, and because of the rather rough approximations (tight-binding), these results cannot be compared quantitatively with the experiment. However, (i) they predict anisotropy energies of the order of 1 to \(2 \times 10^{-3}\) eV/atom, in agreement with more sophisticated calculations [5-7] (exponential values are, however, rather in the range 0.2 to \(1 \times 10^{-3}\) eV/atom); (ii) they clearly show that the anisotropy energy is strongly dependent on the filling of the valence band (the dependence on the valence band filling is due to

<table>
<thead>
<tr>
<th>(N_e)</th>
<th>fcc(001)</th>
<th>fcc(111)</th>
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<tr>
<td></td>
<td>(K_2 [10^{-3}) eV/atom]</td>
<td>(M_I(\mu_B))</td>
</tr>
<tr>
<td>8 (Fe)</td>
<td>-0.00</td>
<td>0.16</td>
</tr>
<tr>
<td>8.5</td>
<td>-1.38</td>
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<tr>
<td>9 (Co)</td>
<td>-1.98</td>
<td>0.33</td>
</tr>
<tr>
<td>9.5</td>
<td>-2.11</td>
<td>0.30</td>
</tr>
<tr>
<td>10 (Ni)</td>
<td>-0.86</td>
<td>0.15</td>
</tr>
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</table>
the fact that the anisotropy is determined mainly by energy states close to the Fermi level and is therefore very sensitive to its position) and (iii) they display the expected anisotropy of the magnetic moment.

2. Effect of the Interface Roughness

Since roughness is always present in real films, it is important to study its influence on the magnetic anisotropy of ultrathin films.

We discuss here the effect of roughness on the dipolar shape anisotropy and on the magnetocrystalline surface anisotropy. We define the roughness by two statistical parameters, the roughness amplitude $\sigma$, which is the mean square deviation from the ideally flat surface, and the correlation length $\xi$, i.e. the average lateral size of flat surface areas.

For our Co films, $\sigma$ and $\xi$ are estimated to be 2 Å and 100 Å respectively [3].

It is well known that for a perfectly flat film, when the magnetization is perpendicular to the film, “magnetic charges” appear on the film sides, resulting in a demagnetizing field $G = -4\pi M$. In the presence of roughness, for in-plane magnetization, some “magnetic charges” appear at the ends of the terraces and craters, giving rise to an in-plane demagnetizing field. Intuitively, the corresponding energy per unit volume is proportional to the roughness-to-thickness ratio, i.e. it is an effective surface anisotropy of dipolar origin. It is also a function of the ratio $\sigma/\xi$. We have performed an analytic calculation of the magnetostatic energy of a film with periodic roughness [4], within the continuous medium approximation (the effect of the discrete nature of magnetization, studied by Draaisma and de Jonge [11], turns out to be small). Apart from the usual dipolar anisotropy

$$E_d = -V2\pi M^2 \sin^2 \theta$$

we obtain the additional term

$$\Delta E_d = 2S2\pi M^2 \frac{3\sigma}{4} \left[ 1 - f\left( \frac{2\pi \sigma}{\xi} \right) \right] \sin^2 \theta,$$

where

$$f(x) = \sum_{n=0}^{\infty} \sum_{m=0}^{\infty} \frac{x^2 \exp(-x^2/4) + x^2 \exp(-x^2/2)}{(2n+1)^2(2m+1)^2}.$$

This dipolar surface anisotropy is always positive. For our Co films, its contribution amounts to less than 0.05 erg cm$^{-2}$, which is much smaller than the measured value. But for very rough this dipolar surface anisotropy may be the dominant contribution.

We have also studied the effect of roughness on the magnetocrystalline surface anisotropy within Néel’s model [10]. In this model [4], the magnetocrystalline surface anisotropy arises from the asymmetric environment of surface atoms as compared to those of the bulk. When some in-plane neighbours are missing (at the ends of terraces) this asymmetry is in some sense reduced. Thus we find that the roughness modifies the magnetocrystalline surface anisotropy by an amount:

$$\Delta K_s = -\frac{2\sigma}{\xi}.$$

For our Co samples, this reduction should be less than 5%, but according to this argument the magnetocrystalline surface anisotropy is strongly affected by a large roughness.

3. Anisotropy Induced by Lattice Mismatch

Because of the magnetostriction, strains in the film induce some anisotropy. An important cause of strains is the lattice mismatch between the film and the substrate. We recently suggested that lattice-mismatch-induced strains can give rise to an effective surface anisotropy (i.e. proportional to $1/t$) [3, 12].

First we present a phenomenological model for interfacial strains. When there is a mismatch $\eta = (a_2 - a_1)/a_1$ between the substrate lattice parameter $a_1$ and the deposit lattice parameter $a_2$ (in the following $|\eta| \ll 1$ will be assumed) this mismatch may be relaxed either by a homogeneous strain of the film or through the introduction of interfacial dislocations [13], as shown in Fig. 1. The exact behaviour is determined by the competition between these two mechanisms. The energy cost of a homogeneous strain is proportional to the volume of the film, whereas the interfacial dislocation energy is proportional to its area, so that we generally expect pseudomorphism at low thicknesses and interfacial dislocations at large thicknesses.

For our Co films, its contribution amounts to less than 0.05 erg cm$^{-2}$, which is much smaller than the measured value. But for very rough this dipolar surface anisotropy may be the dominant contribution.

![Fig. 1. Accommodation of lattice mismatch by interfacial dislocations (squares: substrate atoms; triangles: adsorbate atoms)](image-url)
The strain energy is
\[ E_s = V \frac{1}{2} C \varepsilon^2, \]  
where \( C \) is an elastic constant, \( V \) the volume and \( \varepsilon \) the strain.

The dislocation energy is
\[ E_d = S \mu |1/a - 1/a|, \]  
where \( S \) is the area, \( \mu \) is the dislocation energy per unit length, \( a \sim 1 \) is a numerical factor depending on the dislocation geometry and \( |1/a + 1/a| \), with \( a = a_d(1 + \varepsilon) \), is the dislocation density per unit area. The interaction between dislocations is neglected in (15).

The equilibrium state is obtained straightforward by minimizing the total energy \( E_s + E_d \) with respect to \( \varepsilon \). Below a critical thickness given by:
\[ t_c \sim \frac{\varepsilon \mu}{a_d C |\eta|}, \]  
we have pseudomorphism (i.e. \( \varepsilon = -\eta \)). Above \( t_c \) there are misfit dislocations and
\[ \varepsilon \sim -\eta \frac{t_c}{t}, \]  
as shown in Fig. 2. Although very simple, this model gives a fairly good description of the experimental behaviour: such a variation of \( \varepsilon \) with \( t \) is observed, for example for Ni on Cu [14]. If \( |\eta| \) is too large, \( t_c \) can be smaller than 1 atomic layer and pseudomorphism never occurs.

We can now consider the magnetoelastic energy induced by the strains,
\[ E_{MB} = V B c \sin^2 \theta, \]  
where \( B \) is a magnetoelastic constant. The strains due to the dislocations do not contribute, since their average value is zero. Thus for \( t > t_c \), we have
\[ E_{MB} = SK_s \sin^2 \theta \]  
with:
\[ K_s = -B \eta t_c, \]
so that the anisotropy induced by the mismatch is an effective surface anisotropy. For hcp(0001) Co we have
\[ B = \frac{2c_{13}^2 - (c_{11} + c_{12})k_{33}^3}{c_{33}} (\lambda_A + \lambda_B), \]
where the \( C \)'s and \( \lambda \)'s are, respectively, the usual elastic and magnetostriction coefficients. For Co on Au, \( \eta = -14\% \) and RHEED experiments suggested that \( t_c = 2 \text{Å} \) [15]. This gives a magnetoelastic surface anisotropy of \( K_s = 1.8 \text{er cm}^{-2} \) for the Au/Co interface which is larger than the experimental result. (In [3, 12], the relaxation in the direction normal to the plane was neglected, resulting in a smaller value for \( K_s \).)

Thus, although the situation considered here (film with one free surface) is not exactly the same as in our experiments (sandwich structure) we believe that, besides a possible magnetocrystalline surface anisotropy, our films are also affected by an important contribution due to magnetoelastic surface anisotropy.

The main difference between this mechanism and the magnetocrystalline surface anisotropy is that the anisotropy is distributed throughout the whole film, instead of being concentrated at the interfaces. This would lead to differences in the spin-wave spectrum and therefore in the magnetization distribution at finite temperatures.

\[ \text{Fig. 2. Thickness dependence of the strain in a Ni film grown on Cu(001).} \]

**Theoretical calculation**

- Experimental points [14]; solid line: theoretical result
4. Experiments on Magnetic Surface Anisotropies

4.1. Short Survey

In principle, magnetic surface anisotropy (MSA) can be determined by measuring the magnetic anisotropy as a function of the film thickness, \( t \). For \( t \) values which are not too small (above a few monolayers), the magnetic anisotropy has a \( 1/t \) linear dependence \([\text{relation (2)}]\) whose slope \( 2K_{\text{eff}} \) is the sum of the contributions \( K_1 \) and \( K_2 \) of the two surfaces of the film. This \( 1/t \) dependence was observed e.g. by Gradmann [16, 17]. Since their review paper, published in 1986 [1], the experiments have been extended to many other systems including the pure transition metals Ni, Fe, and Co on various substrates and orientations and alloys such as MnSb [18]. The interesting new developments are related to the cases where MSA favours an easy axis perpendicular to the film plane i.e. \( K_z > 0 \).

The effective volume anisotropy constant is in general negative because of the predominant contribution of the demagnetizing energy \( (-2\pi M^2, \text{for a film with uniform magnetization } M) \). Thus for a critical value of the thickness \( t_c = -2K_{\text{eff}}/K_{\text{eff}} \), the sign of \( K_{\text{eff}} \) changes. This leads to a switching of the spontaneous magnetization from the in-plane orientation for \( t > t_c \), to perpendicular to the film plane for \( t < t_c \). Below the critical thickness, the film can keep a large perpendicular remanent magnetization which exceeds the parallel one.

4.2. Magnetic Surface Anisotropy of Ferromagnetic Transition Metals: Experimental Data

Nickel. The MSA of thin films of fcc Ni(111) deposited on Re(0001) and the changes induced by metal coatings and gas absorptions have been studied in detail by the group of Gradmann [1, 17, 19]. In addition, the MSA of Ni(111) in sputtered Ni/Cu multilayers has been recently investigated [20]. The main results of these studies are the following:

(i) The magnetization of Ni(111) ultrathin films always lies within the plane of the films i.e. \( K_z < 0 \). This is consistent with the absence of perpendicular remanent magnetization in ultrathin Au/Ni/Au(111) sandwiches [21].

(ii) \( |K_z| \), which has the largest value \((0.48 \text{ erg/cm}^2)\) for the free Ni(111) surface, is decreased by about a factor 2 upon coating with non-magnetic metals and is reduced to nearly zero by adsorbed oxygen or NiO coverage [19] (Table 2).

Iron. Because of the existence of a powerful microscopic tool – the \( ^{57}\text{Fe} \) Mössbauer effect – the MSA of ultrathin Fe films has been widely studied in various situations, mainly \( \alpha\)-Fe(100) and (110), and \( \gamma\)-Fe(100).

The nearly perfect \( 1/2 \) relationship between the \( \alpha\)-Fe and Ag lattice leading to a very good match \((+0.8\%)\) when the Fe net is rotated 45° about the surface normal has allowed reliable experiments on epitaxial films and superlattices. For Fe(100)/Ag(100), the easy axis of MSA is normal to the film plane. The critical thickness below which the spontaneous magnetization becomes perpendicular is \( t_c = 2.4 \text{ ML} \) for superlattices [22] and \( 2 \text{ ML} < t_c < 3 \text{ ML} \) for Au/Fe/Ag sandwiches [23], depending on the quality of the Fe/Ag interface. Ferromagnetic resonance studies by the group of Heinrich [24, 25] give a relatively large \( K_z \) value of about \( 1 \text{ erg/cm}^2 \) for the free Fe(100) surface which is noticeably reduced upon coating with the

<table>
<thead>
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<th>System</th>
<th>Temperature [K]</th>
<th>( K_z ) [erg/cm²]</th>
<th>Ref.</th>
</tr>
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<td>UHV/Ni(111)</td>
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<td>[1]</td>
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<td>-0.22</td>
<td>[1, 20]</td>
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<td>Re/Ni(111)</td>
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<td>[1]</td>
</tr>
<tr>
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<td>≈0</td>
<td>[19]</td>
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<td>+1</td>
<td>[24, 25]</td>
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<td>[24]</td>
</tr>
<tr>
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<td>[31, 32]</td>
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<td>[28]</td>
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<td>293</td>
<td>≈0</td>
<td>[41]</td>
</tr>
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noble metals Ag and Au (Table 2). This is consistent with the observation by spin polarized photoemission of a perpendicular remanence of up to 4.5 ML for uncoated Fe(100)/Ag(100) films [26].

The MSA of Fe(110) grown epitaxially on GaAs [27] and W(110) [28] strongly differs from the previous case. Indeed $K_s$ is negative leading to an in-plane magnetization. We shall not discuss the in-plane easy axis switching from the [100] to [110] with decreasing $t$, first observed by the group of Prinz [27] and examined in detail by Rado [29] and Gradmann et al. [28]. We just quote the order of magnitude of the total MSA (2 surfaces) for Fe/W(110) without coverage or covered by noble metals: $K_s \approx -1$ erg/cm$^2$ [30]. This value seems to be more realistic than the value deduced by Rado for Fe(110)/GaAs, $K_s = -3.2$ erg/cm$^2$.

There are some other data on MSA of $\alpha$-Fe. For Fe/Au(100), the surface magneto-optic Kerr studies of Bader et al. [31] show that the magnetization is in the plane of the film along a [100] axis as for bulk $\alpha$-Fe. This was confirmed by spin polarized low energy diffraction [32]. Notice the existence in this system of a segregation of the Au atoms from the substrate to the surface. For Fe(110)/Ag(111) sandwiches, no significant MSA was detected down to 2 ML [33] whereas for Fe(110)/Ag(111) sandwiches, a noticeable MSA leads to perpendicular magnetization for $t < t_c \approx 2$ ML [34]. Summarizing, $\alpha$-Fe thin films show a large variety of MSA since $K_s$ can be either positive, negative or close to zero, depending on the lattice orientation and on the nature of the interface (Table 2).

Iron can also adopt the fcc metastable phase ($\gamma$-Fe) in thin films. The magnetic properties of this phase are extremely sensitive to the lattice constant. Indeed the ground state is antiferromagnetic at small lattice constant and ferromagnetic at larger ones [35]. The epitaxial films of fcc Fe on Cu(100) are ferromagnetic down to one monolayer thickness [36, 37]. Films grown at 100 K are perpendicularly magnetized between 1.5 and 5.7 ML [37]. Assuming 2.2 $\mu_B$ per Fe atom, the value of 5.7 ML for critical thickness should lead to $K_s \approx 1.9$ erg/cm$^2$ for the two surfaces of $\gamma$-Fe/Cu(100) which is about two times larger than the value $K_s \approx 1$ erg/cm$^2$ obtained in Cu/Fe/Cu(100) sandwiches by Brillouin light scattering at room temperature [38].

**Cobalt.** We have studied in detail ultrathin films of hcp Co deposited at room temperature on atomically flat polycrystalline Au(111) and covered by Au. A variety of experiments including ferromagnetic resonance [2], SQUID magnetometry [3] and magnetoresistance [39] clearly show that the spontaneous magnetization of these Au/Co/Au(111) sandwiches becomes perpendicular to the film plane below $t_c \approx 12$ Å (6 ML). So far, this is the largest thickness for which perpendicular magnetization has been observed in coated films. The hysteresis loops obtained at 10 K for different Co thicknesses (Fig. 3) show clearly that the easy axis is perpendicular to the film at low thicknesses. Indeed square hysteresis loops are observed for $t < t_c$. For the thinnest films with $t < t_c$, a striking anomaly of the perpendicular magnetoresistance is observed near the coercive field $H_c$ as shown in Fig. 4 for a double sandwich Au/Co/Au/Co/Au with $t = 7.5$ Å. The large increase of the resistance at $H = 2H_c$ is related to some degree of disorder arising from a distribution of alternately up and down magnetic domains.

An interesting new feature of these very thin Co films with perpendicular magnetization is their large coercivity which exhibits a $t^{-2}$ dependence (Fig. 5). An equally strong thickness dependence seems to hold also for other thin films with perpendicular magnetization such as fcc Fe/Cu [37] and Co/Pd(111) multilayers [40].
Magnetic Surface Anisotropy of Transition Metal Ultrathin Films

**Fig. 4.** Room-temperature perpendicular magnetoresistance of a Au/Co/Au/Co/Au film with Co thicknesses of 7.6 Å and a Au interlayer thickness of 30 Å.

![Graph showing magnetoresistance](image)

**Fig. 5.** Coercive field $H_c$ of Au/Co/Au films, at $T=10$ K, versus $t^{-2}$ ($t$ is the Co thickness).

![Graph showing coercive field](image)

In Table 2 we report the MSA of Co/Au(111) and Co/Cu(111) obtained by FMR in our group and the MSA of Co/Pd(111) measured by Draaisma et al. in Co/Pd multilayers. In contrast to the large perpendicular anisotropy of these films, the magnetization of Co films on Cu(100) remains in the film plane even at monolayer thickness [41].

### 4.3. Effect of Surface Roughness on Magnetic Anisotropy

The theoretical models developed in Sect. 2 show that MSA is strongly affected by roughness of surfaces or interfaces. There is also some experimental evidence for this phenomenon. For instance, the multilayer films are generally less anisotropic than the simple sandwich, and the films obtained by sputtering have a smaller MSA than the films slowly grown in ultrahigh vacuum (UHV). An example is given by Co/Cu(111) for which a large positive MSA is found for UHV grown sandwiches while sputtered multilayers have negligible MSA and thus are magnetized in-plane even at the monolayer thickness [42]. The frequently observed scattering in experimental data on the same system and the deviations from the 1/t dependence of the effective anisotropy could also be related to roughness effects. So far there have only been a few systematic studies of these effects. An interesting way to modify the structure of surfaces or interfaces is to change the growth temperature or to anneal the samples prepared at, or below room temperature. This was done by den Broeder et al. for ion beam sputtered Co/Au(111) multilayers [43]. The samples prepared at room temperature exhibit a MSA which is about two times smaller than that of our UHV-grown Au/Co/Au(111) sandwiches, yielding a critical thickness for perpendicular magnetization of about 5 Å (instead of 12 Å). When annealing the samples at 250–300 °C, the Co/Au interfaces are sharpened because of a segregation of Co and Au atoms and consequently a strong increase of perpendicular anisotropy is obtained.

We have prepared Au/Co/Au sandwiches at room temperature without annealing the Au substrate. In this case, grazing incidence X-ray reflectometry indicates that the Au surface is not flat, in contrast with the one annealed at 150 °C in UHV. The ratios of perpendicular to parallel remanent magnetization for different annealing temperatures clearly reflect the expected increase of perpendicular anisotropy (Table 3). However, the hysteresis loops for perpendicular fields are not as square as the ones obtained by annealing the Au substrate before Co deposition.

### 5. Conclusion

So far, the magnetic surface anisotropy is the most striking phenomenon identified in ultrathin films. Indeed, the MSA energy reaches $10^{-3}$ eV per atom, which is two to three orders of magnitude larger than the bulk value for cubic ferromagnets. In many cases,
this MSA is able to overcome the demagnetizing energy at low thicknesses. This leads to the spectacular result of the spontaneous magnetization turning out of the film plane. An increasing number of experimental data is now available. They are roughly consistent with the theoretical predictions, despite some discrepancies probably due to the many competing mechanisms—like surface and interface imperfections—which can act to modify the MSA.

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