

Spin reorientation transition in Fe(110) thin films: The role of surface anisotropy

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We report on the spin reorientation phase transition in Fe(110) ultrathin films, epitaxially grown on a W(110) single crystal. The critical film thickness of the magnetic reorientation (t_r) has been monitored by spin-resolved photoemission. Either an increase or a decrease of t_r can be induced by controlled modification of the surface (anisotropy) of the Fe films. t_r is reduced by deposition of noble metal (Ag and Au) overlayers on the Fe(110) surface. The effect is significantly more pronounced in the case of Au overlayers than Ag. In contrast, t_r is found to increase by small amounts of oxygen (0.25 monolayer) adsorbed on the Fe(110) surface.

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I. INTRODUCTION

The total free energy of a ferromagnet depends upon the orientation of the magnetization with respect to the crystallographic axis, thereby originating a magnetic anisotropy.

Magnetic anisotropies are of great technological importance: they define easy and hard magnetization axes. The prospect of effectively manipulating magnetic anisotropies is a major drive in applied research. Reorientation transitions occupy a special place in this field, the stability of the magnetization direction being very relevant for data storage applications.

Theoretically, the microscopic mechanisms determining the magnetic anisotropy are relatively well understood. However, detailed computations of specific systems still present a considerable challenge to theory. Magnetic anisotropy energies are typically extremely small (in the order of a few μ eV/atom). Only recently have the *first-principles* calculations approached the level of accuracy required to treat them correctly. It seems therefore appropriate to provide experimental data for comparison with theory.

In ultrathin magnetic films, the subtle interplay between exchange interaction, (long-range) dipolar interaction, and on-site magnetocrystalline anisotropy gives rise to a variety of magnetic phenomena with no counterpart in three-dimensional systems, such as perpendicular magnetic anisotropy and magnetic reorientation transitions.¹ As a result, the study of magnetic properties of thin films has become an important testing ground for understanding the basic interactions and how they are affected by such factors as composition and microscopic structure. This is particularly true for magnetic reorientation transitions, which can be triggered by different factors such as temperature,²⁻⁵ film thickness,^{6,7} and chemical composition. For example, morphology-dependent magnetic anisotropy has been studied in systems such as Fe/Ag(100) (Ref. 8) and Co/Cu;⁹ composition-driven spin reorientation transitions have instead been observed in alloy systems such as Fe_{1-x}Co_x/Cu(100);¹⁰ and surface steps have been reported to be responsible for the magnetic anisotropy in Fe/W(100) system.¹¹

In epitaxial Fe(110) ultrathin films, an *in-plane* to *in-plane* reorientation of the magnetization can be induced by varying the film thickness. This reorientation has been ob-

served when Fe films are grown on such different substrates as GaAs(110) (Ref. 12) or W(110).¹³

The thickness dependence of the reorientation transition in Fe(110) is described in terms of a competition between volume and surface anisotropy contributions to the total energy of the system.¹³ Furthermore, earlier studies have shown that the surface anisotropy term is modified by capping the clean Fe(110) surface with a noble-metal overlayer,^{13,14} effectively altering this equilibrium.

Here we use spin-resolved photoelectron spectroscopy to further investigate the magnetic reorientation transition in the Fe(110)/W(110) system. We report on a systematic study of the variation of the critical switching thickness (t_r) as a function of the thickness of various overlayers. These changes directly reflect a change in the magnetic surface anisotropy of the Fe(110) film. We find that the detailed behavior strongly depends on the specific overlayer used. The three systems analyzed here, Ag/Fe, Au/Fe and O/Fe, exhibit distinctly different dependence of t_r on overlayer coverage.

Ag and Au overlayers on Fe have been chosen as representative of weak electronic interaction between substrate and overlayer, whereas oxidation of the Fe surface represents a case of a stronger interaction. Upon noble-metal deposition, we observe a continuous reduction of the reorientation critical thickness (t_r). The reduction of t_r is considerably more pronounced with Au than Ag overlayers. The case of oxygen adsorption is more complex; at low coverage, t_r increases, attaining a maximum when the surface undergoes an O $c(2 \times 2)$ surface reconstruction. At higher oxygen coverage, t_r decreases and eventually saturates at a value slightly thinner than for the clean Fe surface. These results show that the critical switching thickness t_r in ultrathin Fe films can be manipulated effectively and accurately via modification of their surfaces.

II. EXPERIMENTAL AND DATA PRESENTATION**A. Experiment**

The experiments have been performed at the U5UA undulator beamline at the National Synchrotron Light Source located at the Brookhaven National Laboratory.¹⁵ Spin- and angle-resolved photoemission spectra have been collected with a 50-mm hemispherical analyzer coupled with a low-energy diffuse scattering spin polarimeter. The total energy resolution was set to 100 meV and the angular resolution was

about 2° . All measurements have been performed in ultra-high vacuum (UHV) conditions (base pressure was 5×10^{-11} Torr) at room temperature.

The W(110) surface was prepared following standard procedures described in the literature.¹⁶ A clean W(110) surface—displaying a (1×1) low-energy electron-diffraction (LEED) pattern—was obtained by repeated cycles of annealing ($T_{\text{ann}} = 1800$ K) in an oxygen atmosphere ($p_{\text{O}_2} = 5 \times 10^{-8}$ Torr), followed by a few seconds of flashing at high temperature ($T_{\text{fla}} = 2500$ K) in UHV conditions.

To study the reorientation transition, wedge shaped epitaxial Fe(110) films were grown *in situ* by *e*-beam evaporation. The total pressure rose to about 2×10^{-10} Torr during film depositions. The Fe films evaporated on the W(110) crystal exhibited a sharp (1×1) LEED pattern.¹⁷ The use of wedged samples ensured identical preparation conditions for the various thicknesses. The deposition rate was measured with a quartz-crystal monitor and the corresponding nominal film thickness was calculated using the bulk Fe density and lattice constant.

All magnetic measurements have been performed in remanence. A magnetization pulse (full width at half maximum ~ 0.5 msec, $I_{\text{max}} \sim 100$ Oe) applied in an in-plane direction at 45° between the $[001]$ and $[\bar{1}\bar{1}0]$ crystallographic directions was found sufficient to saturate the sample. Two orthogonal in-plane components of the polarization vector have been measured simultaneously using a two dimensional spin detector.¹⁸

B. Reorientation in pure Fe(110)

Typical results showing the magnetic reorientation transition as monitored with spin-resolved photoemission are reported in Fig. 1. An Fe wedge film, spanning a range of thicknesses across the transition, is grown epitaxially on a W(110) crystal. The valence-band spectra are measured as a function of Fe film thickness along this wedge. In the left (right) panel, the photoemission spectra are decomposed using the spin polarization measured along the $[\bar{1}\bar{1}0]$ ($[001]$) direction. One can observe that the thinnest films (bottom three spectra) are magnetized along the $[\bar{1}\bar{1}0]$ direction, whereas thicker films (top two spectra) are magnetized along the $[001]$ direction. The critical thickness (t_c) is between 80 \AA and 86 \AA for these pure Fe films.

Since we are interested in characterizing the nature of the magnetic reorientation transition, it is convenient to extract from Fig. 1 the information about the spin-polarization vector. This can be accomplished by plotting the average spin polarization as measured in a narrow binding-energy window centered at ~ 4 eV binding energy, below the $3d$ emission. A typical plot of this “secondary” electron spin-polarization is shown in Fig. 2. This plot represents a magnetization profile along the wedge, and allows one to determine precisely the transition thickness. Two things should be noted in Fig. 2. First, the sharpness of the reorientation transition: the entire transition occurs in ≈ 1 ML [$d_{\text{Fe}(110)} = 2 \text{ \AA}$] of thickness change. This is quite surprising considering that the average roughness of these films is certainly greater than 2 \AA at this high Fe film thickness ($\sim 85/2 \sim 42$ ML). Evidently, the mag-

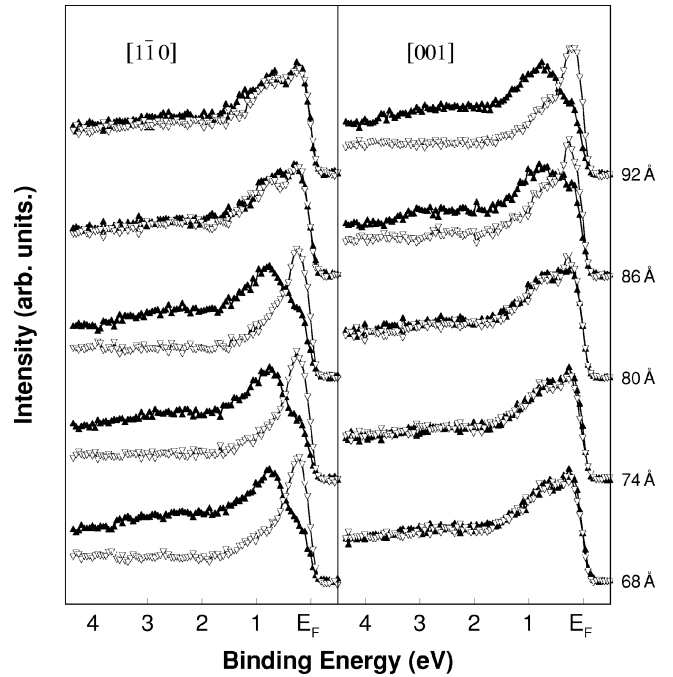


FIG. 1. Spin-resolved valence-band photoemission spectra acquired as a function of Fe thickness along the two in-plane high-symmetry directions. The spectra were measured at 40 eV photon energy, in normal emission. The filled and empty triangles represent majority- and minority-spin components, respectively.

netic coupling efficiently smoothes the magnetic roughness with respect to the structural roughness.

The second observation from Fig. 2 is that the sum of the two spin polarizations remains almost constant across the transition region. This suggests the coexistence of both $[\bar{1}\bar{1}0]$ and $[001]$ domains during the reorientation transition. In the limiting case of a single domain rotating its magnetization from the $[\bar{1}\bar{1}0]$ to the $[001]$ direction, the spin polarization at the crossover point would be $1/\sqrt{2} = 70\%$ of the saturation value, not the observed 50% polarization value displayed in Fig. 2. The existence of two sets of domains is

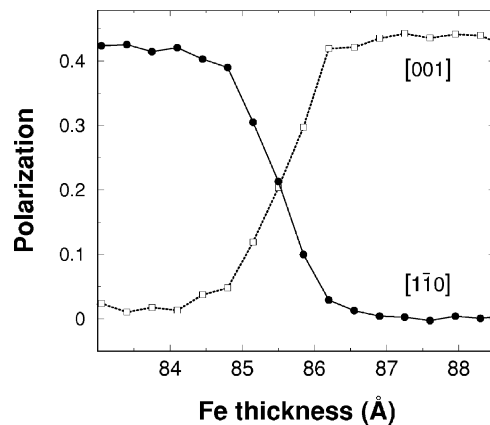


FIG. 2. Spin polarization of “secondary” emission (4 eV binding energy, see Fig. 1) as a function of Fe film thickness.

reasonable considering that the 85 Å thickness is only a nominal value with considerable deviations due to the film roughness. From a microscopic point of view, the crossover thickness (t_r) must correspond to 50% of the crystal surface already thick enough to have the easy axis along the [001] direction and 50% still pinned along the [110] direction.

As mentioned above, the easy axis in ultrathin films is determined by various anisotropy contributions operating simultaneously. In most cases, the dominant terms are the shape anisotropy (representing the dipole interaction) and the magnetocrystalline anisotropy (reflecting the microscopic electronic anisotropy due to the crystal lattice).

The shape anisotropy tends to minimize stray fields, thereby forcing the magnetization to lie in the film plane. Consequently, shape anisotropy does not play any role in the in-plane to in-plane transition of Fe(110) thin films.

As a first approximation, the magnetocrystalline anisotropy can be decomposed into two parts: the volume part and the surface part. The crystal anisotropy is obviously profoundly altered at the surface (and interface). Consequently, the total energy of the system contains a strong surface anisotropy term, independent of the film thickness. In this model, the easy magnetization direction for a given Fe film thickness is determined by a balance between these two terms, the surface anisotropy, which favors the (in-plane) [110] direction, and the bulk anisotropy, which favors the [001] direction. The bulk anisotropy eventually prevails as the film thickness increases and the magnetization switches from the [110] direction to the [001] direction.

Note that in 3d metals, the orbital moment is effectively quenched during the formation of the solid. The bulk component of the magnetocrystalline anisotropy (originating from the spin-orbit interaction) is therefore very small in these systems, considerably smaller than the surface anisotropy. This explains the high t_r value observed here for an in-plane to in-plane reorientation transition [$t_r \sim 50$ ML in Fe/W(110)] as compared to the cases of out-of-plane to in-plane transitions [$t_r \sim 5$ ML in Fe/Cu(100) and Fe/Ag(100)]. In the latter, t_r is determined by the two comparable terms: the shape anisotropy and the surface anisotropy, the bulk anisotropy being negligible in this case.

The simple reorientation model mentioned above for Fe(110) relies on the balance between bulk and surface contributions to the magnetocrystalline anisotropy. It should then be possible to modify this balance by modifying the Fe surface. In the remainder of this work, we follow this idea and consider the modification of the surface of the magnetic material induced by the deposition of a nonmagnetic overlayer.

C. Influencing the Fe reorientation by modifying the surface: Noble-metal overlayers

First, we consider the case of noble-metal overlayers. The choice of noble metals is a convenient way to minimize the interaction with the magnetic (3d) electrons of Fe. Consider, for example, the valence-band spectra shown in Fig. 3. The spectra are measured at various stages during Ag (left panel) and Au (right panel) deposition on the clean Fe(110) surface. The Fe 3d emission (bottom-most spectrum) extends only

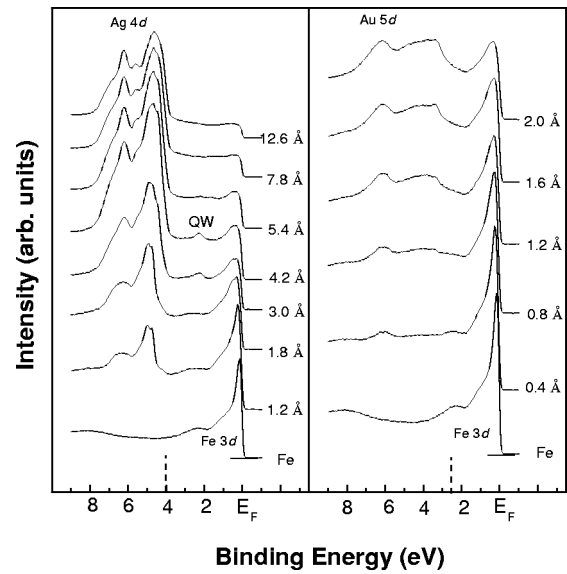


FIG. 3. Valence-band photoemission spectra for Ag (left panel) and Au (right panel) overlayers grown on Fe(110). The spectra are measured at 40 eV photon energy, in normal emission. The thickness of the overlayers is indicated.

about 3 eV below the Fermi level. The emission from Ag 4d and Au 5d states appears during deposition at binding energies greater than 3 eV. In both cases, the d states of the noble metal are well separated from the Fe 3d, and direct d-d interaction is minimal. The effect of the noble-metal overlayers on the magnetic reorientation can be monitored in a similar way as described for the pure Fe. Magnetization profiles from Fe wedged samples have been obtained for various thicknesses of Ag or Au overlayers. The Fe reorientation thickness (t_r) extracted from these data is shown in Fig. 4. The switching transition for both Ag and Au overlayer sys-

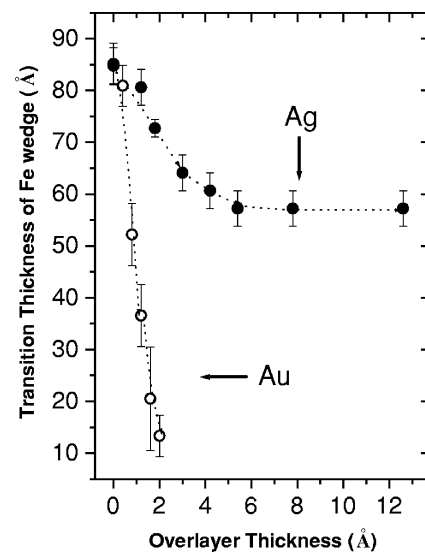


FIG. 4. Fe(110) Critical reorientation thickness (t_r) as a function of the thickness of the noble-metal overlayer. Closed symbols refer to Ag; open symbols refer to Au. The dotted lines are guides for the eye.

tems occurs at a lower thickness (t_r) than that of the clean Fe(110) surface. Furthermore, the effect of the overlayer is considerable. t_r is reduced as much as $\sim 30\%$ for thick Ag films, and much more for Au films.

Qualitatively, this behavior is plausible. The main effect of the weakly interacting noble-metal overlayers is to confine the surface charge of the magnetic layer. The noble metal overlayers transform the topmost Fe layer into a more bulk-like layer. The tendency of the Fe(110) surface to stabilize the magnetization along the $[1\bar{1}0]$ direction is then reduced. Correspondingly, the surface-to-volume anisotropy balance shifts toward thinner films.

Although qualitatively the effect of Ag and Au is the same, i.e., to anticipate the reorientation transition at thinner Fe film thicknesses, quantitatively there are considerable differences between the two cases. Most probably, these differences reflect the different growth modes of Ag and Au on the Fe(110) surface.

For the Ag/Fe(110) wedge surface, the transition thickness decreases linearly to $\sim 70\%$ of the clean Fe(110) surface value after the first 2 ML, and then saturates at higher coverage. Apparently, at room temperature, Ag grows on Fe(110) in close to the ideal layer-by-layer mode. This behavior is consistent with the observation of a sharp hexagonal LEED pattern from this system and the presence of intense Ag *sp*-derived quantum well (QW) states in the valence band of Ag(111)/Fe(110) (see states marked QW in Fig. 3). The saturation effect, reached at only 2 ML of Ag, compares well also with the first-principles calculations¹⁹ which show that the magnetic moment of Fe surface atoms decreases from $2.65\mu_B$ to $2.17\mu_B$ when capped by a Ag monolayer. The $2.17\mu_B$ value is very close to the bulk Fe moment ($2.2\mu_B$).

The case of Au is quite different: t_r decreases much more drastically than Ag. This behavior can be partly attributed to the stronger electronic interaction in Au-Fe than in Ag-Fe systems. The maximum of the Au *5d* band is at ~ 2 eV below the Fermi, while the maximum of the Ag *4d* band is at ~ 4 eV binding energy. However, the fact that no saturation effect takes place with Au would suggest that some other structural mechanism, not present in the Ag overlayers, is operating the Au case. Indeed, the LEED pattern rapidly deteriorates during Au deposition and disappears after only 1 ML of Au. The growth of Au on Fe(110) is much less ordered than the case of Ag, i.e., Au does not grow epitaxially on Fe(110). Moreover, we cannot exclude some intermixing of Au with Fe at the interface. Au forms various stable bulk alloys with Fe. Furthermore, the formation of a Au-Fe surface alloy has been recently observed by scanning tunnel microscope for submonolayer depositions of Au on Fe(100) at temperature higher than 370 K.²⁰ The drastic decrease of t_r reported in Fig. 4 with less than 1-ML Au coverage, in spite of the weak direct *d-d* electronic interaction, is then plausibly related to a profound structural modification of the Fe surface layer.

Before leaving the subject of noble-metal effects, it is useful to compare our results with the current views on magnetic reorientation transitions in ultrathin films. Up to now we have described the thickness dependence of the reorien-

tation as the result of an equilibrium between two different magnetocrystalline anisotropies: the bulk anisotropy, which increases with the film thickness and the surface anisotropy, which is thickness independent. As mentioned above, the model was proposed several years ago in connection with early observations of reorientation transitions in ultrathin magnetic films.^{12,13} Subsequent work, notably on the system Ni/Cu(100),²¹ has emphasized the importance of the magnetoelastic anisotropy in ultrathin films. Epitaxial ultrathin films can form considerably strained structures, due to the lattice mismatch at the interface. Consequently, magnetoelastic terms can become large in these systems and even dominate the magnetic anisotropy terms.

The magnetoelastic terms obviously vary with film thickness, mirroring the process of strain relaxation during film growth. For example, in the specific case of Fe(110) on W(110), there are now elegant measurements directly showing the change of strain in few-monolayer-thick Fe films during growth.²² These results clearly indicate the importance of magnetoelastic anisotropy in this system. Moreover, they prove the incorrectness of assuming bulk anisotropy constants for ultrathin films.

In Refs. 22 and 23, Sander Co-workers also point out that in principle a model considering the thickness dependence of magnetoelastic anisotropy could also explain the observed reorientation in Fe(110) thin films as a function of thickness. This model would not require any surface effect at all. Nevertheless, as also noted in Refs. 22 and 23, the symmetry breaking at the surface still remains a powerful physical argument. Its effect on the magnetic reorientation transition must then be addressed case by case on the basis of experimental results.

Here we have shown that modification of thin-film surfaces is an effective way to shift the critical thickness t_r . Electronically, the deposition of noble-metal overlayers is the least possible modification of the Fe surface. The fact that even in this case t_r varies considerably suggests that the state of this system is indeed determined by an equilibrium between surface and “bulk” energies. Surface effects play an important role in determining the spin reorientation in Fe(110) films.

We conclude that an appropriate model to describe the magnetization in Fe(110) has to take into account magnetoelastic effects as well as surface contributions to the magnetocrystalline anisotropy.

D. Influencing the Fe reorientation by modifying the surface: Oxygen overlayers

The deposition of noble-metal overlayers considerably decreases the critical Fe thickness t_r . Basically, the dominating effect is structural and it simply tends to suppress the surface anisotropy.

More complex situations are expected in cases of stronger electronic interaction between the overlayer and the magnetic surface. In principle, the Fe surface anisotropy could even be enhanced by an appropriate choice of overlayer. The reorientation would then appear at higher Fe thicknesses.

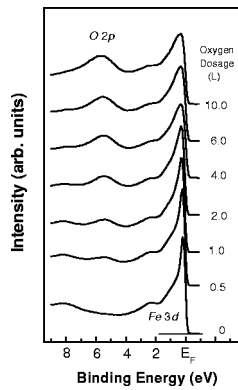


FIG. 5. Valence-band photoemission spectra of Fe(110) as a function of oxygen exposure (in Langmuirs). The spectra are measured at normal emission with 40 eV photon energy.

To test this possibility, we considered the modification of an Fe surface induced by oxygen adsorption, i.e., a case of strong substrate-overlayer interaction. Oxygen on Fe(110) is a well characterized system. In the early stages of oxidation, oxygen adsorbs in a well defined sequence of ordered overlayers: $c(2 \times 2)$ (0.25 ML), followed by $c(3 \times 1)$ (0.33 ML), followed by the formation of a hexagonal FeO(111) layer (1 ML).¹⁷ The presence of two surface reconstructions, with different surface symmetries, makes this system even more interesting for the study of magnetic anisotropies.

The evolution of the valence band as a function of oxygen adsorption is shown in the photoemission spectra of Fig. 5, in the range of oxygen exposures up to 10 L (1 L = 10^{-6} Torr sec).

The O_2 molecules dissociate upon adsorption on the Fe surface. The characteristic O $2p$ structure, located at about 6 eV binding energy, appears in the valence-band spectra. LEED patterns allow the clear identification of the surface reconstructions mentioned above. The most intense $c(2 \times 2)$ LEED pattern is observed at about 1.5 L oxygen adsorption, and the $c(3 \times 1)$ at about 4 L.

As already described for the noble-metal case, t_r is determined using the spin-resolved data from Fe wedged samples for each oxygen exposure. The plot of t_r vs oxygen exposure is shown in Fig. 6.

The behavior of t_r with oxygen is different than that of noble-metal overlayers. For small oxygen exposures, the Fe reorientation thickness increases, and reaches a maximum coincident with the appearance of the $c(2 \times 2)$ ordered overlayer. t_r decreases at higher oxygen adsorption and, by the time the $c(3 \times 1)$ reconstruction occurs, t_r has returned to nearly the same value as for the clean Fe(110) surface. Further oxygen exposure slightly reduces the critical thickness.

The case of oxygen adsorption is interesting: it demonstrates the possibility of inducing significant changes in the Fe reorientation in both directions. Apparently, the Fe surface anisotropy increases at low coverage (<1.5 L) and starts to be suppressed only at higher oxygen exposures. At first sight, it would seem that this behavior cannot be explained by a simple geometrical effect as considered for the noble-metal case. One could suppose that the overlayer-substrate interaction has now become an important factor, at least at low

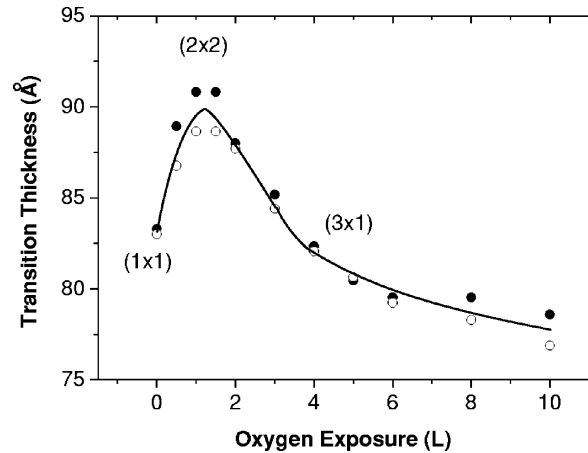


FIG. 6. Dependence of the Fe critical reorientation thickness (t_r) on oxygen exposures.

coverages. The strong electronegativity of oxygen may well be responsible for more profound modifications of the Fe surface electronic structure as compared to the noble-metal case. Enhancements of the Fe surface magnetocrystalline anisotropy could then be associated with the $c(2 \times 2)$ oxygen overlayer. A realistic calculation of the electronic structure and total energy for this specific system could possibly shed some light on this point. Inspired by these findings, we came across an interesting piece of experimental work on a subtle structural effect induced by oxygen in a related system: it has been observed that submonolayer adsorption of oxygen on 2-ML $p(1 \times 1)$ Fe/Ag(100) induces an out-of-plane to in-plane reorientation of the magnetization.²⁴ The important point for the present work is that this magnetic reorientation is associated with a structural change of the Fe film. The adsorbed oxygen seems to act as a sort of (post) surfactant. The as-deposited 2 ML Fe(100) film contains a considerable density of three-dimensional islands. The oxygen promotes a reordering of the Fe film in which it becomes flatter, as indicated by electron-diffraction data.²⁴

Inspired by these findings, we monitored the evolution of the LEED spots from the Fe(110) films during the oxygen uptake. Typical results of these measurements are shown in Fig. 7. The spot size shrinks during the first 1.5 L oxygen exposure, i.e., during the formation of the $c(2 \times 2)$ oxygen overlayer. This effect saturates above 1.5-ML oxygen exposure, with the spot size remaining practically unchanged upon further oxygen exposure.

The width of the LEED spots is related to the density of three-dimensional islands on the surface. Their decrease in size indicates a smoothening of the surface. We conclude that a postsurfactant effect, similar to the one seen in the Fe(100) films, is operating for the Fe(110) films also. This observation suggests a different interpretation of the t_r -dependence from oxygen adsorption. As mentioned above for the case of pure Fe, t_r represents a nominal thickness, i.e., an average thickness taken over an appropriate distribution of Fe islands of different heights. The LEED observations show that the oxygen adsorption effectively narrows the island size distri-

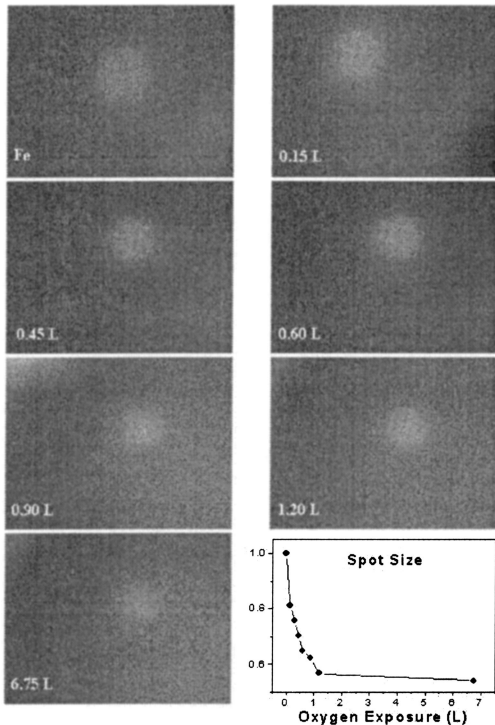


FIG. 7. Evolution of the $\bar{[11]}$ LEED spot profile from 6-ML Fe(110) on W(110) during oxidation. Oxygen exposures were performed at room temperature at an oxygen partial pressure $p=5 \times 10^{-9}$ Torr. Oxygen doses are indicated in Langmuir (1 L = 1×10^{-6} Torr sec).

bution, i.e., diminishes the roughness of the surface. This effect can have consequences for the magnetic reorientation transition.

The first consequence of smoothening is that Fe atoms on the higher peaks are redistributed to fill the lower valleys. With considerable Fe roughness, this process initially tends to reinforce the thinner regions of the sample, i.e., the one favoring the $[110]$ easy magnetization direction, at the expense of the thicker regions which favor the $[100]$ direction. The balance between these two is then altered in favor of the former. We have to move to a region of the wedge of higher (average) thickness to find the magnetic reorientation.

A second mechanism is operating in ultrathin films due to the presence of step edges. Atoms at step edges have reduced symmetry with respect to atoms on a surface terrace. Bruno has presented a general theoretical argument that this circumstance always leads to a decrease of the surface anisotropy in the presence of roughness.²⁵ This means that in our case a decrease of the roughness increases the surface anisotropy and consequently even the thick regions of the sample will be less prone to orient along the $[001]$ direction. Once again, the balance between thin and thick regions is altered in such a way that the reorientation transition is shifted to higher thickness.

In summary, the experiment with oxygen shows that the reorientation transition is not defined by the average thickness of the Fe(110) film alone. A more complete description of the transition requires consideration of the roughness of the film, i.e., at least the second moment of island size distribution.

Finally, above 1.5 L exposure, the structural transformation of the O/Fe surface is complete. Apparently, at this point the oxygen overlayer has a similar effect as Ag and Au. The critical thickness decreases by $\sim 20\%$ as the oxygen exposure increases above the $c(2 \times 2)$ overlayer (i.e., >1.5 L), a value comparable to the decrease observed in Ag.

III. CONCLUSIONS

In summary, in-plane reorientation of the magnetic easy axis in ultrathin Fe(110) films has been studied by spin-polarized photoemission. Considering the reorientation as the result of a balance between surface and volume magnetocrystalline anisotropy energies, we have investigated the modification of this equilibrium induced by the deposition of various overlayers. The use of films with wedge geometry is particularly convenient for this purpose. It allows us to monitor accurately the shift of the critical thickness t_r as a function of the overlayer deposition. In all three cases analyzed, Ag/Fe, Au/Fe, and O/Fe, we have found a considerable effect on the reorientation thickness t_r . This emphasizes the key role played by the surface in this reorientation.

Beyond this general consideration, the three specific systems studied display different behavior, each shedding some light on different aspects of the magnetic reorientation transition.

Ag represents the simplest case. Its effect on the Fe switching transition is limited to the first two monolayers deposited and consists of a reduction of $\sim 30\%$ in the critical thickness t_r . This suggests a suppression of the Fe surface anisotropy.

For the Au overlayer system, the reduction in t_r is much more dramatic. Less than 1-ML equivalent of Au shifts the critical thickness by more than 80%. We argue that the formation of a composite material at the interface must be responsible for this behavior.

Finally, the case of oxygen overlayers is even more varied. Both an increase in t_r at low oxygen exposures (<1.5 L) and a decrease at high exposures (1.5–10 L) are observed. The perturbation of the surface electronic structure of Fe induced by the oxygen overlayer is certainly more complex than that associated with the noble-metals. Therefore, its role in this effect must be important and should be addressed by a proper theoretical investigation. However, the coincidence of the increase in t_r with the reordering of the film surface, suggests the following interpretation of the shape of the t_r curve vs oxygen exposure. At low oxygen dosage (<1.5 L), t_r increases as a result of the smoothening of the film surface. At higher exposures, the suppression of the surface anisotropy becomes the dominant effect and t_r decreases, similarly to what observed in the noble overlayer cases.

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