

Epitaxial fcc Fe–Co alloy films on Cu(001)

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Abstract

Fcc Fe–Co alloy films (1.5–9 monolayers) were prepared on Cu(001) by epitaxial deposition at room temperature. Growth, structural and magnetic properties of these films were investigated by medium- and low-energy electron diffraction, Auger electron spectroscopy, as well as magneto-optical Kerr effect measurements. Addition of Fe in the fcc Co films results in an uneven vertical expansion of a part of the alloy films, its extent depending on composition. For more than 70 at.% Fe, essentially the whole film is vertically expanded, which is accompanied by a change of the growth mode and the appearance of some uniaxial structural irregularity in the plane of the films. The easy axis of the magnetization was found to lie within the film plane over the whole range of compositions for the investigated thicknesses. A linear increase of the Kerr signal at saturation magnetization with increasing thickness indicates that the whole alloy film is magnetic. A constancy of the Fe and Co local magnetic moments can be expected below 70–80 at.% of Fe.

Keywords: Epitaxy; Alloys; Structural properties; Magnetic properties and measurements

The 3d transition metal alloys and in particular Fe–Co compounds currently receive considerable interest in materials science due to promising technological applications. Both bulk and thin film Fe–Co samples have been taken under consideration. Whereas structural, magnetic, and electronic properties of the former are rather well understood [1,2], the parameters of the thin films often differ from those in the bulk, and can to a great extent be influenced by using different deposition techniques and various substrates [3–5]. In this paper we present the results of a multi-technique investigation of epitaxial FeCo films on the Cu(001) surface. The motivation for the choice of the substrate was two-fold. On the one hand, a possible pseudomorphic growth of the alloy films on the fcc Cu(001) allows us to stabilize the fcc phase of these alloys (stable in the bulk above 1 260 K) over a wide range of compositions. On the other hand, growth, magnetic and structural properties as well as the electronic structure of both Fe and Co ultrathin films on Cu(001) have been rather extensively studied (see, for example, Refs. [6–10]) and can be used as a basis to analyze the same parameters for the alloy films.

The $\text{Fe}_x\text{Co}_{1-x}$ films were epitaxially deposited in a UHV chamber by coevaporation from two separate Knudsen cells onto the Cu(001) single crystal kept at room temperature. The deposition rate was selected between 0.3 and 1.0 monolayers (ML) per minute, the thickness of the films varied from 1.5 to 9 ML, and the entire region of concentrations was covered with steps of 10–20%. The growth process was mon-

itored and controlled in situ by medium-energy electron diffraction (MEED) and the resulting composition of the films was determined with an accuracy of 5% by Auger electron spectroscopy (AES) using calibration curves based on the comparison of MEED and AES data for the individual Fe and Co films on Cu(001).

During the codeposition of Fe and Co, the diffracted beams in the MEED experiment show pronounced oscillations which are characteristic for the film composition (Fig. 1). Up to a relative Fe concentration of 60–70%, these oscillations are very similar to those for deposition of only Co onto Cu(001) [10]. They persist up to the maximum thickness investigated (9 ML), indicating a nearly perfect layer-by-layer growth of the alloy films on Cu(001). For higher Fe concentrations only the first three or four oscillations were observed, which points towards a deviation from the layer-by-layer growth mode or some structural rearrangements in the films. Interdiffusion of Co and Fe with Cu during the deposition and following measurements at room or lower temperatures can be practically excluded, because the ratio between the low-energy Auger peaks of Co (53 eV) and Cu (60 eV), which is very sensitive to such an interdiffusion, did not change up to 430 K for the thinnest films (1.5 ML) investigated.

Conclusions on the crystalline structure of the films are drawn from the interpretation of low-energy electron diffraction (LEED) patterns and the analysis of LEED $I(E)$ curves for the (00) diffraction beam. The LEED patterns for the

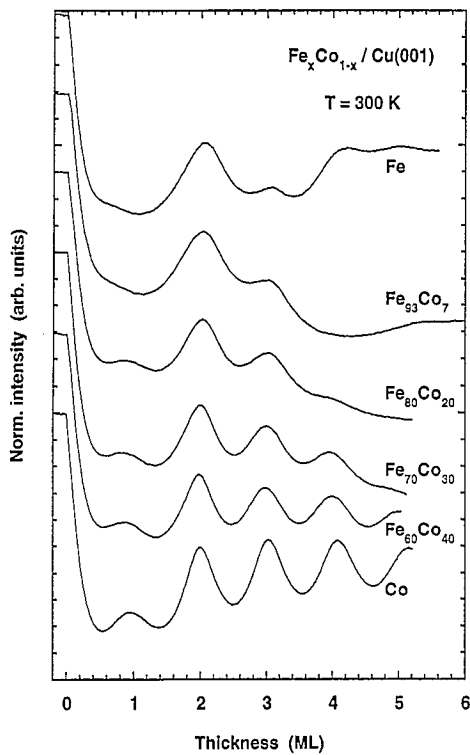


Fig. 1. Evolution of the MEED specular intensity during deposition of $\text{Fe}_x\text{Co}_{1-x}$ on Cu(001) for various values of x . The curves are normalized to the intensity before deposition. The primary beam ($E_{\text{kin}}=2$ keV) was directed along the [100] azimuth with an polar angle of incidence of 3–5°.

FeCo alloy films (measured at room temperature) reveal diffraction spots at the same positions as the fcc spots of clean Cu(001) over the whole range of compositions and thicknesses investigated, thus indicating a pseudomorphic growth of the alloys. The spots are sharp for the films with Fe concentrations below $\approx 70\%$ (Fig. 2(a)), but become rather diffuse and elongated for the Fe-rich alloy films (Fig. 2(b)): the (10) and equivalent spots look more like stripes (see Fig. 2(c), where the corresponding profile is depicted), whereas the (11) spots (not shown) become cross-like. The broadening of the spots may be attributed to a change in morphology of the films due to a deviation from the layer-by-layer growth mode, whereas their elongation may be explained by the appearance of some uniaxial (in the [110] direction for one domain) structural irregularity in the plane of the films. A fcc–bcc structural transition as a reason for the elongation seems less probable, because the main spots do not disappear (as they do, for example, at this transition for Fe on Cu(001) [8]), but only change their form.

More information can be obtained from the analysis of the LEED $I(E)$ curves of the (00) diffraction beam for films with thicknesses of 5 ML (Fig. 3). Such curves often show a periodic sequence of intensity maxima (as a function of \sqrt{E}) that results from constructive interference at coincidence of the vertical interlayer spacing with an integer number of the electron wavelength [11]. The $I(E)$ curves in Fig. 3 reveal two such peak sequences with relative weights depending on the composition: at $x=0$ and $x=0.95$ only one of these structures is observed, whereas both of them are

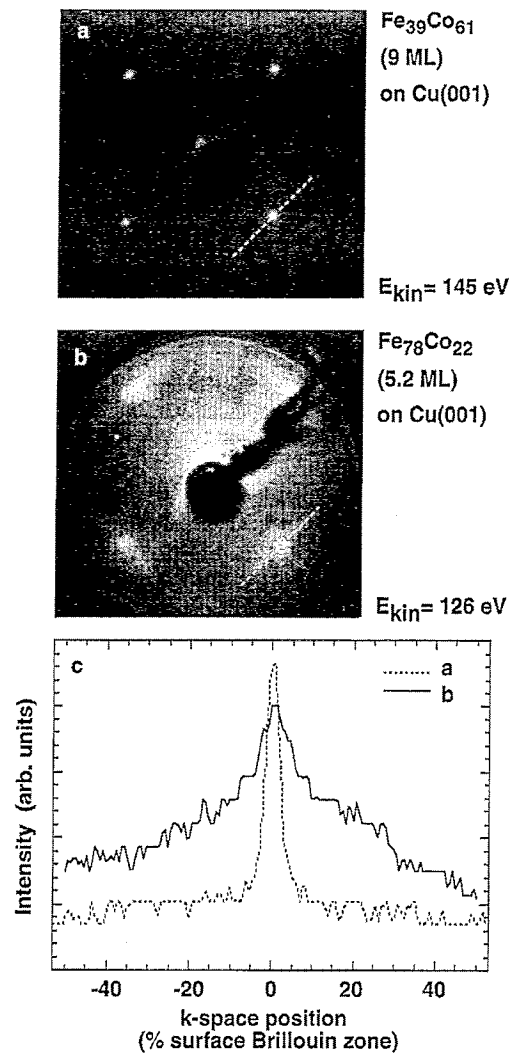


Fig. 2. LEED patterns for 9 ML of $\text{Fe}_{39}\text{Co}_{61}$ (a) and for 5.2 ML of $\text{Fe}_{78}\text{Co}_{22}$ (b) on Cu(001). Intensity profiles of the (10) spot along the white dashed lines in the LEED patterns (c).

present in the intermediate concentration regime. It indicates the presence of two different interlayer distances in the investigated films; the corresponding values (evaluated in the kinematic approximation) are depicted in the insert of Fig. 3 as a function of composition.

Both the absolute values of these distances and their continuous development from those in the fcc Co and Fe films define the structure of the FeCo alloy films as a vertically modified fcc structure (fct) in complete agreement with the character of the LEED patterns. In fcc Fe films on Cu(001) a coexistence of two different vertical interlayer distances (a contracted and an expanded fcc phase) has been previously found [6,8], and is clearly reflected in the $I(E)$ curve for Fe in Fig. 3. The interlayer distance in the Fe-rich alloy film corresponds to the distance of the expanded phase in the Fe film. Pure Co films on Cu(001) have a tetragonally compressed fcc structure with a single vertical interlayer spacing [9]. Addition of Fe to the fcc Co film not only results in a continuous increase of this spacing but also leads to an uneven vertical expansion of a part (presumably, some lateral

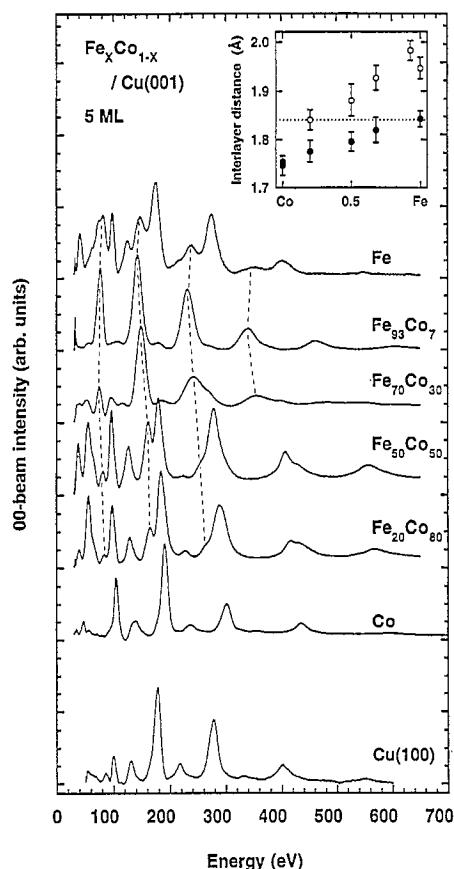


Fig. 3. Intensity vs. voltage dependence of the LEED (00) beam for Cu(001) (bottom) and 5 ML $\text{Fe}_x\text{Co}_{1-x}$ on Cu(001) at various values of x . The curves were taken with a polar angle of incidence of 6° in the (100) mirror plane of the substrate. The second periodic sequence of intensity maxima appearing with increasing x is traced by dashed lines. Two different interlayer distances (full and open symbols, respectively), calculated from the depicted $I(E)$ curves are presented in the insert as a function of x . The respective value for clean Cu(001) is marked with a dashed line.

regions) of the alloy film, its extent depending on the composition. The corresponding periodical sequence (traced by dashed lines in Fig. 3) is present only at low energies in the Co-rich alloy films. It may therefore be associated with low-energy electrons having a small escape length, which supposes that the uneven expansion probably starts at the surface in small patches, spreading then both laterally and to the subsurface layers. For more than 70 at.% Fe practically the whole film is expanded, which is accompanied by the change of growth mode and the appearance of some disorder in the film. For completely expanded fcc Fe films on Cu(001) (at coverages below 4 ML) a regular parallel shift of the adjacent rows in the [110] direction appears, resulting in the (4×1) or (5×1) superstructures [6]. Analogous irregular shifts can cause the additional LEED structure for Fe-rich alloys. Considering that the expanded layers of the fcc Fe films on Cu(001) are responsible for the ferromagnetism of these films [6,7], one would expect that the Fe rich alloy films are also completely ferromagnetic.

The magnetic properties of the Fe–Co films were characterized by recording hysteresis loops in polar and longitudinal geometries using magneto-optical Kerr effect (MOKE)

measurements. The easy axis of magnetization was always found to lie within the film plane over the whole range of thicknesses and compositions investigated. Although the easy axis in the fcc Fe films on Cu(001) is perpendicular to the film plane, we did not observe a magnetic signal in this direction for the Fe-richest alloy films investigated (93% Fe), at least in the range of magnetic fields accessible ($H_{\text{max}} = 250$ Oe). It is known that the magnetic anisotropy in a thin film contains contributions from shape, magnetocrystalline, strain, and surface anisotropies. On the one hand, strain is believed to be the origin of the perpendicular magnetization in fcc Fe on Cu(001) [6]. On the other hand, a magneto-elastic anisotropy contribution due to the fct lattice distortion in Co films on Cu(100) has been made responsible for the in-plane magnetization in the system [12]. In this case shape and strain anisotropies act in the same direction, and cannot be compensated by the surface contribution at any Co film thickness. This can explain the results for the Co-rich alloy compositions. For the Fe-rich films the situation is more complicated. Data for the magneto-crystalline anisotropy in the bcc FeCo alloys alters from [111] to [100] for more than 60% Fe. It should be noted that in our films we did not observe a pronounced in-plane anisotropy, although some differences between hysteresis loops for the [100] and [110] directions, which are the only low-indexed directions within a {100} crystallographic plane, can be found. Whereas for thicker films of the Fe-rich alloys certainly the shape anisotropy favours an in-plane magnetization, we tentatively attribute the absence of a perpendicular magnetization in the monolayer regime to a complex interplay of surface, magneto-crystalline, and magneto-elastic anisotropies.

The in-plane saturation magnetization for $\text{Fe}_{50}\text{Co}_{50}$ films on Cu(001) reveals a steady increase with increasing film thickness from 2 to 9 ML (see Fig. 4(a)). This shows that, at least at this composition, the whole film is magnetic, except maybe some small amount of Co and Fe at the interface. This conclusion is also supported by the observation that the Curie temperature T_c in the $\text{Fe}_{50}\text{Co}_{50}$ film continuously increases with the coverage, the values of T_c being rather similar to those in fcc Co on Cu(001) [10]. In particular, for 2.5 ML of $\text{Fe}_{50}\text{Co}_{50}$ T_c reaches the temperature corresponding to the onset of the interdiffusion (≈ 430 K) and thus its measurement at the higher coverages is impossible. Looking at the influence of the composition on the magnetization, one finds an approximately linear increase with the Fe concentration for all experimental points, apart from one at 88% Fe (Fig. 4(b)). The Fe and Co magnetic moments are known to be different ($M_{\text{Co}} = 1.7 \mu_B$; $M_{\text{Fe}} = 2.25 \mu_B$ in the bulk) and a linear dependence on the concentration should be expected if these moments were constant over the whole composition range. A deviation from this dependence would thus indicate a change of the magnetic moments, or may also correspond to the loss of long-range ferromagnetic order in a part of the film. The latter is, in principle, possible for the Fe-rich films

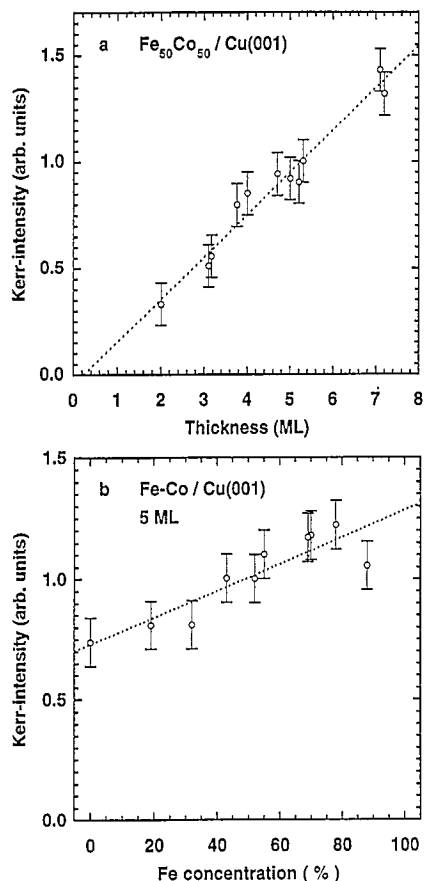


Fig. 4. Saturation magnetization (in the [100] direction) for the hysteresis loops measured using magneto-optical Kerr effect versus thickness for the $\text{Fe}_{50}\text{Co}_{50}$ films on Cu(001) (a) and versus relative Fe concentration for 5 ML $\text{Fe}_x\text{Co}_{1-x}$ films on Cu(001) (b). The measurements were carried out at room temperature which is significantly less than T_c in the films (T_c lies slightly above room temperature only for 2 ML of $\text{Fe}_{50}\text{Co}_{50}$). Dashed lines represent linear fits.

if some amount of the contracted fcc phase is still present; at the higher Fe concentration this phase could become antiferromagnetic as in Fe on Cu(001). As to the possible change of the magnetic moments, in bulk bcc Fe–Co alloys an increase of the magnetic moment of Fe from $2.25 \mu_B$ to $2.6 \mu_B$ with addition of Co takes place up to a Co concentration of about 30% [2], causing a non-monotonous compositional dependence of the magnetization with a maximum at this concentration [1]. Bulk bcc Fe is known to have an incompletely filled majority spin band (weak ferromagnetism), and the addition of Co results in an effective intra-atomic charge transfer which leads to a complete filling of the majority bands

(strong ferromagnetism). On the other hand, for disordered fcc Fe–Co monolayers on Cu(001) a stabilization of strong ferromagnetism even at the Fe-rich end of the alloy due to the pinning of the Fermi energy to that of the Cu substrate and the narrowing of the overlayer density of states in a monolayer film has been predicted [13]. This stabilization would imply a nearly constant magnetic moment of Fe over the whole compositional range and consequently a linear increase of the total overlayer magnetization with increasing Fe content. Unfortunately, the current lack of the experimental information for the Fe-rich alloys films does not allow us to make an unequivocal conclusion yet, whether or not the weak or strong ferromagnetism takes place at the high Fe concentration. Further measurements are necessary.

Acknowledgements

This work has been supported by BMFT under grant No. 05 5EFAAI 5.

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