Chapter 3

Epitaxial growth and magnetic coupling across Fe_{50}Mn_{50}

The experimental results on the magnetic interlayer coupling across an AFM spacer layer, Fe_{50}Mn_{50} in the present studies, are grouped together in this chapter. The characterization of the layer-by-layer growth and the surface morphology of Fe_{50}Mn_{50} on Cu(001) are presented in Sec. 3.1. In Sec. 3.2 layer-resolved domain imaging and MOKE measurements on FeNi/Fe_{50}Mn_{50}/Co and Co/Fe_{50}Mn_{50}/Co trilayers outline the influence of surface roughness in the phase and strength of coupling. Sec. 3.3 and Sec. 3.4 present the interlayer coupling in Ni/Fe_{50}Mn_{50}/Co and Ni/Fe_{50}Mn_{50}/Ni trilayers on Cu(001). The latter leads to some conclusions on the magnetic spin structure of Fe_{50}Mn_{50}.

3.1 Growth and surface morphology of Fe_{50}Mn_{50} alloy on Cu(001)

Magnetic properties such as interlayer exchange coupling, giant magnetoresistance, surface anisotropy, are closely related to the surface and interface morphology. It has been shown by Heinrich et al. that the strength of bilinear exchange coupling between Fe and Cr layers can be changed by as much as a factor of 5 by varying the substrate temperature during growth [71]. This behavior revealed clearly how the growth and surface morphology of the FM/AFM interface can influence the magnetic coupling behavior. Therefore, to gain deeper insight into magnetic interlayer coupling across Fe_{50}Mn_{50} alloy thin films, the growth and morphology aspects were investigated first.

As a bulk material FeMn alloy exhibits a variety of different structural phases, depending on composition. In the Fe rich part of the phase diagram a supersaturated hcp solid solution ε has been found, while in the Mn rich part a solid solution of α-Mn is present. The Fe_{50}Mn_{50} alloy as a bulk material has an fcc γ-FeMn structure. The stabilization of the antiferromagnetic fcc γ-phase is of particular importance especially as a possible application in spin valves [72],
The bcc $\alpha$-phase and hcp $\varepsilon$-phase, both present in the phase diagram of $\text{Fe}_{1-x}\text{Mn}_x$ alloys, have a lower ordering temperature.

Ideal layer-by-layer growth in heteroepitaxy is generally believed to require a small lattice mismatch, immiscibility in the bulk, and an abrupt interface between substrate and film. For FeMn these requirements can be met by Cu(001) [73]. The (001) surface of Cu is a very popular substrate for growth of metallic thin films. Cu exhibits an $fcc$ crystal structure and an equilibrium lattice constant $a_{\text{Cu}} = 3.61 \text{ Å}$ i.e., the misfit to $\gamma$-FeMn ($a_{\text{FeMn}} = 3.629 \text{ Å}$) [74] amounts to $f = \frac{a_{\text{FeMn}} - a_{\text{Cu}}}{a_{\text{Cu}}} = 0.52\%$.

Figure 3.1: RHEED specular beam intensity oscillations vs time acquired during deposition of $\text{Fe}_{50}\text{Mn}_{50}$ alloy on Cu(001) as a substrate. The inset shows the RHEED pattern of 10 ML $\text{Fe}_{50}\text{Mn}_{50}$/Cu(001).

The growth of the films was monitored in situ by a RHEED experiment at grazing incidence. Figure 3.1 presents the RHEED specular spot intensity acquired during the growth experiment. The substrate temperature during deposition was held at 300 K, and the evaporation rate was about 0.5 ML/min. The inset of Fig. 3.1 shows the RHEED pattern for an incidence angle of 0.7°–0.8° (first anti-Bragg condition) and an azimuthal direction near the $\langle 110 \rangle$ Cu crystallographic axis after deposition of 10 ML $\text{Fe}_{50}\text{Mn}_{50}$. The presence of the RHEED oscillations is a strong indication for a layer-by-layer growth of $\text{Fe}_{50}\text{Mn}_{50}$ on Cu(001). The initial phase of the growth is characterized by a pronounced decrease of intensity and an intensity minimum at 2 ML $\text{Fe}_{50}\text{Mn}_{50}$. Additionally, visual inspection of the RHEED pattern reveals that after deposition of approximately 1 ML $\text{Fe}_{50}\text{Mn}_{50}$ this pattern becomes very diffuse. Connected to
the rise in intensity of the RHEED beams around 3 ML is an increase in spot sharpness. Above this thickness a regular intensity oscillation and an increase in sharpness of the diffraction spots have been observed. From these observations it is obvious that the initial growth mode of the Fe\textsubscript{50}Mn\textsubscript{50} films is not a “perfect” layer-by-layer growth. The absence of regular intensity oscillations in the thickness range of 0–3 ML and the broadening of the diffraction spots are inconsistent with such a growth mode.

The morphology of Fe\textsubscript{50}Mn\textsubscript{50} on Cu(001) has been studied for different coverages using scanning tunnelling microscopy. By depositing low coverages I will first address the quality of the bottom interface, between the Fe\textsubscript{50}Mn\textsubscript{50} alloy and Cu(001). STM images have been taken in constant current mode, at room temperature. The sequence of STM images, together with selected line profiles, of different amounts of Fe\textsubscript{50}Mn\textsubscript{50} deposited at room temperature on Cu(001) substrate are presented in Fig. 3.2.

After deposition of 1.3 ML Fe\textsubscript{50}Mn\textsubscript{50} alloy, Fig. 3.2 (a), the morphology of the surface is dominated by relatively small brighter dots corresponding to the nucleation of islands in the second layer, with an average density of 44 \times 10^3 \mu m^{-2}, seemingly randomly distributed on top of a nearly closed first layer. Mean island sizes are typically of the order of 2–5 nm linear dimensions. A closer inspection of this image shows that while the first monolayer is not fully completed the nucleation and growth of the second one already becomes significant.
These islands resemble basically two different classes (labelled A and B in Fig. 3.2 (a)): (A) with an apparent height of 0.208 nm and (B) with an apparent height of 0.355 nm at $U_{\text{sample}} = 0.5\text{ V}$. The apparent height of islands of (B) type is not the double height of islands of (A) type. Therefore, it must be that A and B islands are of different constituent elements. The effect of imaging the (B), (A) type islands with a lower/higher apparent height is presumably due to a chemical contrast. Variation of the bias voltage shows a significant influence on the apparent height. For a negative bias voltage on the sample of about $-0.6\text{ V}$ the apparent height of (A) type islands is 0.168 nm while of (B) type islands it is 0.303 nm. It can be seen that the apparent height of these islands depends on the bias voltage. A similar bias-dependent contrast has been observed previously in the Cu/W(110) system by Mo and Himpsel [75]. They attributed it to a Cu induced empty state at 0.6 eV above the Fermi level. Although band structure knowledge of the elements is needed to understand the bias-dependence corrugations in detail, it seems reasonable to attribute them to Cu. It may be that the corrugation reflects the diffusion of Cu through the surface of Fe$_{50}$Mn$_{50}$ alloy. Based upon the existing data it is not possible, however, to define a concise structural model. Further investigations with suitable methods (angle resolved ultraviolet photoemission spectroscopy (ARUPS), scanning tunnelling spectroscopy (STS)) should be applied for a full understanding.

In summary, at 1.3 ML there are exposed first monolayer, some holes that correspond to the substrate level and islands nucleated on top of the first monolayer which are one or two monolayers high. The two monolayers high islands (B) reflect the diffusion of Cu from substrate through the first FeMn layer, which act as a pinning centers for the further deposited film.

Deposition of 2.7 ML Fe$_{50}$Mn$_{50}$ alloy onto Cu(001), Fig. 3.2 (b), leads to bigger islands, with the island edges preferentially oriented along $\langle 100 \rangle$ crystallographic directions of the Cu(001) surface. Also, the density of “blobs” (brighter spots), islands nucleated in the next layer (in the third one), has substantially decreased, whereas the apparent width is not much affected and stays at a value of 5 nm.

Based on the STM results described above, the following initial growth mechanism could be proposed. Below 2 ML FeMn, growth proceeds via formation of two dimensional monolayer or bilayer islands in the typical size of a few nm. According to STM observations, the presence of these small islands promotes a decrease of the sharpness of the RHEED pattern and of the specular spot intensity in the 0–3 ML FeMn thickness range.

The formation of small islands in the first stage of growth in collaboration with the absence of RHEED oscillations has been observed in other metal-on-metal systems. Co on Cu(001), for example, grows initially in bilayer islands, followed by layer-by-layer growth [76]. The growth in this low coverage regime is mainly governed by surface free energy effects. It is known that the surface free energy of Co ($\gamma_{\text{Co}} = 2.7\text{ Jm}^{-2}$) is higher than that of Cu ($\gamma_{\text{Cu}} = 1.9\text{ Jm}^{-2}$). It follows then from Bauer’s criteria for the dimensionality of the critical nucleus that the Co atoms will preferably agglomerate in the initial stage of growth. Thermodynamically, exposing
more of the substrate reduces the energy of the system.

The growth of Co on Au(111) also forms small islands two atomic layers high that coalesce with increasing coverage [77]. Here, bilayer growth has been attributed to the large lattice mismatch between Co and Au (14%). The mismatch leads to substantial strain energy in the growing film. The atoms relax in the second layer, thus bilayer islands reduce the overall strain energy of the film. The mismatch between Fe₅₀Mn₅₀ and Cu(001) is much less (~0.52%). In addition, analysis of the RHEED pattern and previous LEED experiments performed on this system indicates that the film remains pseudomorphic with the substrate up to a thickness of 26 ML Fe₅₀Mn₅₀ [78]. Therefore, stress relaxation as a driving mechanism for such a bilayer growth could not be involved in the case of epitaxial growth of Fe₅₀Mn₅₀ on Cu(001).

An alternative explanation relies on the different adatom mobility between the first and successive Fe₅₀Mn₅₀ layers. The difference is responsible for the transition from island nucleation to layer-by-layer growth after 3 ML as observed by STM. Since the surface is covered by a high density of small islands, a significant fraction of the incident atoms will arrive on top of the islands prior to coalescence. Atoms that adsorb on top of the islands may be hindered from descending onto the Cu(001) surface by the presence of a step edge barrier [79, 80]. This additional barrier leads to a build up of adatoms in the second layer prior to the completion of the first. Once the Cu(001) substrate is covered, Fe₅₀Mn₅₀ adatom mobility is increased. This increase produces a sharp transition between completion of the second atomic layer and growth of the third one.

The regular oscillations of the RHEED specular spot intensity observed above 3 ML Fe₅₀Mn₅₀ are regarded as the fingerprint of layer-by-layer growth. Further, the morphology of the Fe₅₀Mn₅₀ alloys in a thickness range of 8 ML to 12 ML will be presented. The sequence of STM topographic images and selected line profiles for different coverages of Fe₅₀Mn₅₀ alloy layers are shown in Fig. 3.3. The STM topographic image of 8.26 ML Fe₅₀Mn₅₀, Fig. 3.3 (a), reveals large and atomically flat terraces. The brighter islands with an apparent height of 0.18 nm correspond to the 9\textsuperscript{th} atomic layer. Small holes (≈ 5 nm length) of one atomic layer depth are also present (darker features). In addition, a fine structure (intermediate gray level) with an apparent corrugation height of about 0.05 nm is recognized. This can be distinguished on the line profile displayed beneath the topography image as a modulation with a small amplitude. Upon increasing the thickness of deposited Fe₅₀Mn₅₀, Fig. 3.3 (a)–(c), the STM topographic images corresponding to 9.15 ML, 10.1 ML and 11.74 ML show mainly the same morphology. Hence, also STM confirms the nearly perfect layer-by-layer growth as already concluded from the presence of RHEED oscillations.

The high features 0.18 nm are steps/islands one atom high/deep. The question is then what is the fine structure seen in all the surface topography images of Fe₅₀Mn₅₀ in Fig. 3.3. The first tendency is to attribute the fine features observed in all STM topographic images to a real geometric effect. The atomic radius of Mn is 5 pm larger than of Fe and could account for
Figure 3.3: Sequence of STM constant current images measured at room temperature: (a) 8.26 ML Fe<sub>50</sub>Mn<sub>50</sub>; (b) 9.15 ML Fe<sub>50</sub>Mn<sub>50</sub>; (c) 10.04 ML Fe<sub>50</sub>Mn<sub>50</sub>; (d) 11.74 ML Fe<sub>50</sub>Mn<sub>50</sub>. Line profiles taken along the width of images, at positions marked by arrows, are shown at the bottom side of each image. Tunnelling current $I = 0.2$ nA, sample voltage $U_{sample} = 0.5$ V.
3.1 Growth and surface morphology of Fe$_{50}$Mn$_{50}$ alloy on Cu(001)

such effects. This implies that the tip retracts 5 pm when it is over a Mn atom. Therefore, the small features present in the topographic images with an apparent step height of 0.05 nm can not be attributed only to geometric effects. To identify the origin of the fine structures observed in the STM topographic images, scanning tunnelling spectroscopy (STS) measurements were performed. Fig. 3.4 (a) and (b) show current map images measured on the same area for positive and negative bias voltages. Fig. 3.4 (c) shows a constant-current topographic image of the same surface area as shown in panels (a) and (b). It is immediately evident from comparison of images (a) and (b) that the gray scale level of the related features changes with the sample bias voltage. Fig. 3.4 shows I–V curves acquired on the entire regions (A) and (B) marked in Fig. 3.4 (a). Qualitatively, in the (A) region the tunnelling I–V curve shows a plateau within 0–0.5 eV, followed by a linear increase of the current. In the negative region the tunnelling curve exhibits a strong exponential increase with a peak in dI/dV at −0.5 eV. When the tip is located over the (B) region, the tunnelling current shows a linear dependence for positive values of the bias voltage, while for negative bias voltage an exponential increase with a peak in dI/dV at −0.2 eV is observed. Since the tunnelling current for a positive bias voltage is higher (lower) above the B (A) region, in constant-current-mode the tip has to be withdrawn from (approached towards) the surface in order to keep the tunnelling current constant. Consequently, the apparent height is reduced for (A) and enhanced for (B) features.

In Fig. 3.5 line profiles taken in the panels (a), (b), and (c) of Fig. 3.4, at the same sample position, show a clear correlation between the features seen in the current map images and the small fine structure in the topography images. The curves labelled (a) and (b) in Fig. 3.5 correspond to the line profiles taken in the current map images for positive (0.4 V) and negative (−0.15 V) voltages, respectively. The (c) curve is a line profile taken in the topography constant current image. The vertical dotted lines are put at some selected positions to guide the eyes. When following the curve labelled (a) in Fig. 3.5, from left to right, some current peaks can be distinguished. They are correlated with a higher tunnelling current, for a positive voltage, when the tip is located over areas which appear brighter in Fig. 3.4 (a). At the same position on the sample, a negative voltage gives rise to a negative peak in the tunnelling current, darker areas in Fig. 3.4 (b). A closer look to the peaks seen in the line profile (a) and (b), Fig. 3.5, reveals that not only the height but also the width of these peaks is slightly different for positive and negative voltages. Now, since the topography image is measured at a positive bias voltage, in the light of the above discussion of Fig. 3.4, a current peak in curve (a) should induce a peak in the topography line profile (c). Indeed, this is what can be observed when the curves (a) and (c) are compared with each other. Most of the small features with 0.05 nm apparent corrugation height are related to features in the current map images.

The simultaneously measured topographic and spectroscopic images confirm clearly the chemical origin of the small features observed in the STM topographic images. The presence of the fine structures may be therefore associated with local concentration differences of the
Figure 3.4: (a) Current map image at 0.4 V of 11.74 ML Fe\textsubscript{50}Mn\textsubscript{50} epitaxially grown on Cu(001). (b) Current map image at −0.15 V, of the same area as in panel (a). (c) STM constant-current topographic image of the same surface area shown in panel (a) and (b). Tunnelling current $I = 0.2$ nA, sample voltage $U_{sample} = 0.5$ V. (d) Averaged I–V curves on the entire areas labelled (A) and (B) measured in panel (a).
Figure 3.5: Line profiles taken in the current map images (a), (b) and topography image (c), in Fig. 3.4, at the positions marked by arrows.

constituents of the chemically disordered alloy. It means that the surface of Fe$_{50}$Mn$_{50}$ shows regions enriched in one or the other of the alloy components. A quantitative interpretation of STM/STS data presented in this section requires knowledge of the electronic surface states of sample and tip. Further experiments with surface sensitive techniques such as ultraviolet emission spectroscopy (UPS), or electron energy loss spectroscopy (EELS) must be applied for a full understanding.

### 3.2 The FeNi/Fe$_{50}$Mn$_{50}$/Co trilayer system

Using photoelectron emission microscopy (PEEM) combined with X-ray magnetic circular dichroism, the magnetic domain structure in coupled epitaxially grown FeNi/FeMn/Co/Cu(001) double wedge-shaped trilayers was investigated. The results have been backed up with magneto-optical Kerr effect measurements of Co/FeMn/Co/Cu(001) trilayers.

Figure 3.6 presents element resolved domain images obtained at the Co $L_3$ edge (left hand side) and the Fe $L_3$ edge (right hand side) of 6 ML FeNi deposited as a continuous film on a 20 ML Fe$_{50}$Mn$_{50}$/8 ML Co crossed double wedge on Cu(001). The Co thickness varies in the range of Fig. 3.6 from 3 to 8 ML from bottom to top, as indicated at the left axis, and stays constant with 8 ML above. The Fe$_{50}$Mn$_{50}$ thickness increases from 0 to 14 ML from left to right, as indicated at the bottom axis. Different gray scales correspond to different projections of the local magnetization direction onto the direction of the incoming light. The crystallographic axes and the direction of the light incidence are indicated by small arrows in the right upper,
In the Co domain images different gray scale levels are recognized corresponding to domains of four different directions of magnetization. Analysis of the contrast reveals that for a thickness below 6 ML Fe\textsubscript{50}Mn\textsubscript{50}, the Co magnetization direction is pointing along \langle110\rangle crystal axes, as indicated by arrows. Above 6 ML Fe\textsubscript{50}Mn\textsubscript{50} thickness a change in the orientation of the magnetization direction of Co towards \langle100\rangle is observed, as indicated by arrows. In the

![Layer-resolved domain images of Co (left) and Fe (right) of 6 ML FeNi/0–14 ML Fe\textsubscript{50}Mn\textsubscript{50}/3–8 ML Co grown epitaxially on Cu(001). Co and Fe\textsubscript{50}Mn\textsubscript{50} thicknesses are indicated at the left and bottom axes of the images, respectively. The direction of the incoming X-rays (h\nu), crystallographic axes, and magnetization direction of some of the domains are indicated by arrows in each figure.](image)

measurements of Kuch et al. a change in the easy axis of Co films on Cu(001) by 45° has been observed when in contact with antiferromagnetic FeMn films [15]. The change of the Co magnetization direction from \langle110\rangle to \langle100\rangle has been related to the transition from paramagnetism to antiferromagnetism of Fe\textsubscript{50}Mn\textsubscript{50} films. In analogy to the fluctuation mechanism proposed by
3.2 The FeNi/Fe$_{50}$Mn$_{50}$/Co trilayer system

Slonczewski, the above result has been explained by an effective 45° magnetic coupling of Co to Fe$_{50}$Mn$_{50}$. Therefore, in accord with these results, in this work it will be assumed that the change in the magnetization direction of Co is the result of the magnetic phase transition of Fe$_{50}$Mn$_{50}$ from paramagnetic to antiferromagnetic for a thickness of about 6 ML.

Inspection of the Fe domain image in Fig. 3.6, corresponding to the FeNi layer, shows that the leftmost one third of the image, for Fe$_{50}$Mn$_{50}$ thicknesses below 3 ML, exhibits an identical domain pattern as the Co image. The analysis of the different contrast levels reveals that for an Fe$_{50}$Mn$_{50}$ thickness below 3 ML the FeNi magnetization is aligned with the Co magnetization direction as shown by arrows. It has been reported previously by Offi et al. that when Fe$_{50}$Mn$_{50}$ is deposited on Co, below a thickness of about 3 ML, it becomes fully ferromagnetic by the proximity to the ferromagnetic Co layer [81]. This ferromagnetic phase leads to a direct exchange coupling between the Co and FeNi layers. The stronger in-plane anisotropy of Co compared to the FeNi anisotropy probably aligns the FeNi magnetization into the same direction as the Co magnetization.

Above 3 ML Fe$_{50}$Mn$_{50}$ thickness an undulation of the magnetization direction of FeNi between the (110) and (100) crystallographic axis is observed as a continuously changing contrast in the FeNi image between 3 and 6 ML FeMn thickness. A continuous change in the contrast from bright to light dark and then to light bright is seen when following the bright stripe-like domain in the Fe image from left to right. In this thickness range the FeNi magnetization is non-collinearly aligned to the Co magnetization. The fourfold in-plane anisotropy of the 6 ML FeNi favors the orientation of the FeNi magnetization along the ⟨100⟩ directions. The interlayer magnetic coupling tries to align the FeNi magnetization along the same direction with the Co magnetization. The competition between these two terms, anisotropy energy of the FeNi and interlayer magnetic coupling energy, leads to a canted configuration.

Increasing the Fe$_{50}$Mn$_{50}$ thickness beyond 6 ML, a periodic oscillatory changing of the contrast of the Fe domain image with a period of 2 ML Fe$_{50}$Mn$_{50}$ thickness is observed. Analysis of the gray scale of the Fe image shows that in this region FeNi and Co magnetizations are either parallel or antiparallel along ⟨100⟩ directions. These oscillations are consistent with layer-by-layer thickness dependent reversals in the orientation of the magnetization of the top FeNi layer, implying that the Fe$_{50}$Mn$_{50}$ layers are antiferromagnetically ordered.

Besides the magnetization oscillations with a 2 ML period of the Fe$_{50}$Mn$_{50}$, additional magnetization changes are observed in the bottom part of the FeNi domain image, Fig. 3.6. A closer look into these stripe-like domains reveals a ripple-like periodic modulation of the coupling phase with a period of 1 ML as a function of Co thickness. On the Co plateau no such undulation of the stripes is seen. This may indicate that for the layer-by-layer growth of Co the change of the roughness when going from a filled atomic layer to a half filled layer induces such a ripple-like modulation of the phase of the coupling. In the last chapter of this thesis a comprehensive discussion will be dedicated to this subject.
Chapter 3. Epitaxial growth and magnetic coupling across Fe\textsubscript{50}Mn\textsubscript{50} thin films

The as-grown PEEM magnetic domain images give information about the direction of the magnetization. They do not lead to a value of the strength of the coupling. The strength of the interlayer exchange coupling, i.e., the coupling energy per unit area, can be determined by applying a varying magnetic field to the trilayer structure and measuring the magnetization curves by conventional magnetometry. Such magnetization measurements have been obtained using magneto-optic Kerr effect (MOKE) at room temperature. The MOKE measurements have been performed on Co/Fe\textsubscript{50}Mn\textsubscript{50}/Co trilayers deposited on Cu(001). The Fe\textsubscript{50}Mn\textsubscript{50} was deposited as a wedge-shaped layer of linearly increasing thickness by moving the sample in front of a shutter during deposition. Here, I underline that the LMOKE measurements have been performed in a second UHV system (MBE-machine) described in Sec. 2.3. Fig. 3.7 (a) shows hysteresis loops of 8 ML Co/Cu(001) (the narrow loop) and a 13 ML Fe\textsubscript{50}Mn\textsubscript{50}/8 ML Co bilayer on Cu(001). The external applied field was aligned along the ⟨100⟩ crystal axes (the magnetic easy axis (EA) of the Fe\textsubscript{50}Mn\textsubscript{50}/Co/Cu(001) system). Note a significantly enhanced coercivity in the coupled film ($H_{c,FeMn/Co} = 33.7$ mT) which is roughly 56 times the coercivity of the single Co layer ($H_{c,Co} = 0.6$ mT) [82, 83]. Figure 3.7 (b) illustrates a typical hysteresis loop, measured in the LMOKE geometry with s-polarization, corresponding to FM coupling between the Co layers. A kink in the loop is seen when we follow the loop from zero value of the magnetic field towards negative values. The kink feature could be associated with the switching of one layer (most likely the upper one), or with 90° orientation of the Co layers. To get more insight into the origin of this kink I have measured MOKE in the longitudinal geometry but with p-polarization. In this case besides the longitudinal component of the magnetization the Kerr signal is also sensitive to the transversal component. Fig. 3.7 (d) shows the MOKE loop measured in the longitudinal geometry with p-polarization, at the same position of the sample. When the external magnetic field is reduced from positive saturation towards zero and then to negative values, an increase (jump) of the Kerr signal can be seen, followed by a plateau and a decrease towards negative saturation. This peak from Fig. 3.7 (d) is located at exactly the same position as the kink in Fig. 3.7 (b). For the reason that the longitudinal component of the Kerr signal has to be monotonic, the peak in Fig. 3.7 (d) must be due to a transverse component.

Since the magnetization reversal is accomplished by an increase of the Kerr signal (Fig. 3.7 (d)) I propose that the kink feature observed in Fig. 3.7 (b) is due to rotation, together, of both layers, to a direction orthogonal to the external magnetic field. The metal films are thin enough so that the Kerr signal is sensitive to both FM layers. An independent rotation of the Co layers would give rise to a second kink in the hysteresis loop of Fig. 3.7 (d), which has not been observed. In other words, the magnetization reversal occurs in two distinct stages: first, by a switching by 90° followed by a second switching towards a direction parallel with the applied magnetic field. Hence, the “shelflike” feature seen in Fig. 3.7 (b) near zero Kerr signal corresponds to the range of fields for which the magnetization vectors are in an intermediate state at $\approx 90°$ to the external magnetic field direction. A closer inspection of the hysteresis loop in Fig. 3.7 (b) reveals...
3.2 The FeNi/Fe$_{50}$Mn$_{50}$/Co trilayer system

Figure 3.7: LMOKE hysteresis loops measured along the ⟨100⟩ crystal axes at RT of: (a) shows two curves corresponding to 8 ML Co/Cu(001) and 13 ML Fe$_{50}$Mn$_{50}$/8 ML Co/Cu(001), respectively. The narrow loop in panel (a) corresponds to 8 ML Co/Cu(001); (b) 6 ML Co/13 ML Fe$_{50}$Mn$_{50}$/8 ML Co measured with s-polarization of the incident beam; (c) 6 ML Co/14 ML Fe$_{50}$Mn$_{50}$/8 ML Co with s-polarization of the incident beam; (d) 6 ML Co/13 ML Fe$_{50}$Mn$_{50}$/8 ML Co with p-polarization of the incident beam; (e) 6 ML Co/14 ML Fe$_{50}$Mn$_{50}$/8 ML Co with p-polarization of the incident beam.
that the “shelflike” features on the positive and negative branches are slightly vertically shifted towards each other. In addition the height of the upper and lower peaks seen in Fig. 3.7 (d) is slightly different. These observations reflect a small misalignment ($\approx 5^\circ$) of the crystallographic [100] axis with the longitudinal measurement direction.

In Fig. 3.7 (c) and (e) are shown hysteresis loops measured in the region of AFM coupling in between the Co layers, $t_{FeMn} = 14$ ML. For large external fields the interlayer coupling is overcome, and the layers are parallel to the field. As the field is reduced, the interlayer coupling exceeds the Zeeman energy and the thicker FM layer starts to reverse. In the remanent state both layers are oriented antiparallel (in a direction orthogonal to the external applied field direction). When the external magnetic field is increased from zero to either positive or negative values, a slight increase (decrease) of the Kerr signal is observed for the measurement with p-polarized light (Fig. 3.7 (e)). It corresponds to the rotation starting point of one FM layer while the second one is fixed.

**Figure 3.8:** Map of the interlayer magnetic coupling of Co/Fe$_{50}$Mn$_{50}$/Co trilayers deposited as continuous films on Cu(001). The black (white) bullets denote a FM (AFM) interlayer coupling, as measured with MOKE. The errors of the thickness estimation are attached to each bullets as vertical (horizontal) bars.

In addition MOKE measurements on Co/Fe$_{50}$Mn$_{50}$/Co trilayers, deposited as continuous films, were carried out. The thicknesses of the Co bottom layer and Fe$_{50}$Mn$_{50}$ were changed in the submonolayer range. Fig. 3.8 presents a cumulative map of these MOKE measurements. The white (black) bullets are associated with antiferromagnetic (ferromagnetic) coupling in between the FM Co layers. It does not contain any information on the strength of the coupling. The Co bottom layer thickness varies from 8 to 8.5 ML, as indicated at the bottom axis, and the Fe$_{50}$Mn$_{50}$ thickness from 12 to 14 ML as indicated at the left axis. The thickness of the Co top
layer is maintained constant at 6 ML for all measurements. For a thickness of the Co bottom layer of 8 ML, and 13.5 ML Fe\textsubscript{50}Mn\textsubscript{50}, the interlayer coupling between the two Co layers is ferromagnetic (FM), as seen also in Fig. 3.7. Increasing the Fe\textsubscript{50}Mn\textsubscript{50} thickness to 14 ML while maintaining the Co bottom layer constant, the interlayer coupling changes to antiferromagnetic (AFM). For a Co bottom layer thickness of 8.5 ML, and an Fe\textsubscript{50}Mn\textsubscript{50} thickness of 14 ML, the interlayer coupling is FM. Hence, a change of the Co bottom layer thickness of 0.5 ML induces a change of the coupling from AFM to FM.

![Figure 3.9](image_url)

**Figure 3.9:** (a) Dependence of the remanence magnetization ($M_r$) normalized to the saturation magnetization ($M_s$) as a function of the Fe\textsubscript{50}Mn\textsubscript{50} spacer layer estimated from LMOKE measurements of Co/Fe\textsubscript{50}Mn\textsubscript{50}/Co trilayer on Cu(001). The Fe\textsubscript{50}Mn\textsubscript{50} spacer layer was deposited as a wedge within a thickness range of 13–14.8 ML. The Co bottom layer thickness was 8.0 and 8.5 ML, respectively, while the Co top layer was similar for the both measurements, 6 ML. (b) Coupling strength ($J_1$) of the same sample as in the case (a) as a function of Fe\textsubscript{50}Mn\textsubscript{50} thickness calculated using eq. 3.1 and the experimental estimated $H_S$ values.

Fig. 3.9 (a) shows the dependence of the remanence magnetization ($M_r$) normalized to the saturation magnetization ($M_s$) as a function of Fe\textsubscript{50}Mn\textsubscript{50} thickness for two different coverages of
the Co bottom layer, 8.0 ML and 8.5 ML, respectively. The top layer is a 6 ML Co continuous film in both cases. The Fe\textsubscript{50}Mn\textsubscript{50} layer was prepared as a 13 ML continuous film followed by the deposition of 2 ML on top as a wedge. Let us first discuss the case of 8.0 ML Co bottom layer thickness (open bullets). Going from 13 ML to 14 ML Fe\textsubscript{50}Mn\textsubscript{50} thickness the normalized remanence magnetization changes from unity to nearly zero. This corresponds to a change of the type of the coupling from FM ($M_r = 1$) to AFM ($M_r = 0$) between the Co layers. Above 14.2 ML Fe\textsubscript{50}Mn\textsubscript{50} the value of remanence magnetization rises up again. Now, for the case of 8.5 ML Co bottom layer thickness, $M_r/M_s \simeq 1$ at 13 ML Fe\textsubscript{50}Mn\textsubscript{50} and decreases to $M_r/M_s \simeq 0$ at only 13.4 ML Fe\textsubscript{50}Mn\textsubscript{50} and increases again to $M_r/M_s \simeq 1$ at 14 ML Fe\textsubscript{50}Mn\textsubscript{50} thickness. Comparing the graphs for the two cases, the $M_r/M_s \simeq 0$ region for 8.0 ML Co bottom layer appears wider than for 8.5 ML Co bottom layer thickness. Since the data reproduce the experimental results measured on two different samples, prepared in the same conditions, a slight change of the evaporation rate during wedge deposition can induce a change of the slope of the wedge. In Fig. 3.9 the errors of the Fe\textsubscript{50}Mn\textsubscript{50} thickness estimation are within 0.2 ML.

It is possible to determine the coupling constant $J_1$ and the equilibrium angles between the FM layers by numerically fitting the MOKE curves following the path of minimum energy as the external magnetic field $H$ is varied [84]. As an alternative in simple systems, some features may be determined analytically. Neglecting the anisotropy of the films, the following explicit relation between $J_1$ and $H_s$, the field at which both layers switch from antiparallel to parallel, can be written:

$$H_s = -\frac{J_1}{t_1 M_1 + t_2 M_2}$$

where $J_1$ is the total bilinear coupling strength and $M_1$, $M_2$ are the saturation magnetizations of layers 1 and 2 with thicknesses $t_1$ and $t_2$. For the numerical evaluation of $J_1$, literature values were chosen for the saturation magnetization of Co/Cu(001) estimated to be about 1.789 T in Ref. [85] and interplanar spacing distance for Co/Cu(001) to be about 1.774 Å in Ref. [86]. The coupling strength values $J_1$ calculated as described above are plotted in Fig. 3.9 (b) as a function of Fe\textsubscript{50}Mn\textsubscript{50} thickness. The change of the Co bottom layer thickness from 8.0 ML to 8.5 ML, for the same Fe\textsubscript{50}Mn\textsubscript{50} thickness, decreases the interlayer coupling strength by nearly 50% from 0.75 mJ/m$^2$ to 0.33 mJ/m$^2$.

Without going into details and possible mechanisms in this section, the phase and the strength of coupling of single crystalline films was seen to be strongly dependent on the submonolayer coverage of the bottom FM layer. In the last chapter of the thesis dedicated to the discussion I will give a qualitative model, and discuss and compare the coupling energy values determined here with the coupling measured in other systems.
3.3 The Ni/Fe\textsubscript{50}Mn\textsubscript{50}/Co trilayer system

This section presents a layer-resolved photoelectron emission microscopy (PEEM) study of the interlayer coupling in Ni/Fe\textsubscript{50}Mn\textsubscript{50}/Co trilayers, epitaxially grown on Cu(001). Epitaxial Co and Ni films on Cu(001) exhibit different magnetic easy axes. Whereas Co films are always magnetized in the film plane, Ni films show a spin reorientation transition from an easy axis parallel to the film plane at film thicknesses below 8 ML to an easy axis perpendicular to the film plane at thickness between 8 ML to 56 ML [89, 90, 91]. For a trilayer in the absence of interlayer coupling, the magnetic easy axes in each magnetic layer are defined by separate minimization of the anisotropy energy. In the case of a strong interlayer coupling, both magnetizations are forced to be collinear, whereas for weak coupling the individual magnetization directions will be along their own easy axes. Reorientation transitions between these two situations can occur for intermediate values of the coupling strength [93]. The coupling across Fe\textsubscript{50}Mn\textsubscript{50} between Co with an in-plane easy axis and Ni with an out-of-plane easy axis will be addressed in this section.

Fig. 3.10 shows element selective domain images obtained at the Co $L_3$ edge (left) and at the Ni $L_3$ edge (right) of 15 ML Ni deposited as a continuous film on an Fe\textsubscript{50}Mn\textsubscript{50}/Co crossed double wedge on Cu(001) as a substrate. The Co thickness varies in the range of Fig. 3.10 from 3 to 8 ML from bottom to top, as indicated at the left axis, and the Fe\textsubscript{50}Mn\textsubscript{50} thickness from 0 to 14 ML from left to right, as indicated at the bottom axis. The crystallographic axes and the direction of incoming light are indicated in the right upper and left bottom part of each panel, respectively. The Co image shows relatively large magnetic domains. When following the image from left to the right, a change in the gray scale contrast at about 7 ML Fe\textsubscript{50}Mn\textsubscript{50} is observed. It reflects a change of the direction of magnetization of Co from $[\bar{1}10]$ towards $[0\bar{1}0]$ above 7 ML Fe\textsubscript{50}Mn\textsubscript{50}. As has been discussed in the previous section, this change in the easy axis of Co after deposition of Fe\textsubscript{50}Mn\textsubscript{50} reflects a phase transition from paramagnetism to antiferromagnetism of Fe\textsubscript{50}Mn\textsubscript{50}.

Inspection of the Ni image of Fig. 3.10 (right) reveals an identical domain pattern as the Co image for an Fe\textsubscript{50}Mn\textsubscript{50} thickness below 3 ML. In this area the Ni magnetization is in-plane, aligned with the Co magnetization direction. A reorientation of the Ni magnetization direction from in-plane to nearly out-of-plane starts at about 3 ML Fe\textsubscript{50}Mn\textsubscript{50}.

Within the thickness range 3 to 7 ML Fe\textsubscript{50}Mn\textsubscript{50} can be seen. A closer look to the dark stripe-like domain at around 5 ML Fe\textsubscript{50}Mn\textsubscript{50} thickness reveals a continuous change of the contrast from light gray at the edge to darker in the middle of the stripe. This can be explained by a gradual canting of the Ni magnetization direction as a function of Fe\textsubscript{50}Mn\textsubscript{50} thickness, probably due to RKKY-like interlayer coupling across the paramagnetic Fe\textsubscript{50}Mn\textsubscript{50} layer as in previous section.

Now, let us follow the Ni image, Fig. 3.10, as a function of the Co bottom layer thickness.
An oscillating behavior when looking to the white stripes of the Ni domain image at 6 ML Fe$_{50}$Mn$_{50}$, from bottom to top of the image, is observed. For a fixed Fe$_{50}$Mn$_{50}$ thickness, increasing (decreasing) the Co thickness by less than one monolayer changes the direction of the Ni magnetization from pointing “up” to pointing “down”. A quite interesting observation is that these oscillatory stripes are present only on the Co wedge, and just above 4 ML Co. On the Co plateau, the upper part of the Ni image, domains of an irregular shape, with an out-of-plane magnetization can be seen. The 1 ML Co bottom layer thickness periodicity of these stripes, and the presence only on the wedge, may indicate a connection with the interface roughness.

There must be a mechanism that leads to a correlation between the as-grown out-of-plane magnetization direction of the Ni layer and the in-plane direction of the underlying Co layer.
Oscillations of the direction of the out-of-plane magnetization could consequently also be related to oscillations of the sign of the interlayer coupling. These could be seen as a clear indication that either the coupling or this mechanism has to do with the roughness of the Co–Fe\textsubscript{50}Mn\textsubscript{50} interface.

Kuch \textit{et al.} have found a correlation between the Ni out-of-plane domains and the Co in-plane domains, in Co/Cu/Ni trilayer system. The Co magnetic domains show different directions of magnetization on top of “up” or “down” Ni domains \cite{92}. However, in Fig. 3.10, the Co bottom layer show a single domain within the white stripes in Ni top layer.

In the Co image the change of the magnetization direction from [\textbar 1\textbar 1\textbar 0] to [0\textbar 1\textbar 0], as mentioned above, is seen for an Fe\textsubscript{50}Mn\textsubscript{50} thickness of 7 ML, exactly the same thickness where the small domains in the Ni layer appear. This thickness thus corresponds to a change of Fe\textsubscript{50}Mn\textsubscript{50} from paramagnetic to antiferromagnetic at room temperature. Similar observations were made by Kuch \textit{et al.} in the as-grown Co on Fe\textsubscript{50}Mn\textsubscript{50} wedge on Cu(001) \cite{15} in this system. Above 10 ML Fe\textsubscript{50}Mn\textsubscript{50} small magnetic domains are formed in the Co layer. The 10 ML corresponds to the thickness of ordering transition at room temperature of the antiferromagnetic Fe\textsubscript{50}Mn\textsubscript{50}. Local exchange interaction between Co and Fe\textsubscript{50}Mn\textsubscript{50} induces a replicate domain pattern of the Fe\textsubscript{50}Mn\textsubscript{50} domains in the Co as-grown layer. Heating the sample above the Néel temperature of Fe\textsubscript{50}Mn\textsubscript{50} promotes larger domains in the Co layer. This may be an indication that the presence of small domains in the as-grown state of a FM layer deposited on top of an AFM layer is related to the local exchange interaction FM/AFM.

![Figure 3.11: Layer-resolved magnetic domain images of a 15 ML Ni/9.8–16.2 ML Fe\textsubscript{50}Mn\textsubscript{50}/8 ML Co trilayer on Cu(001). The domain image of the Co layer is shown on the left hand side, the domain image of the Ni layer on the right hand side. The change of the Co magnetization direction is replicated in the Ni faint stripes domain pattern, superimposed on a domain pattern of small domains.](image)
On the upper right part of the Ni image (Fig. 3.10), above 7 ML Fe\textsubscript{50}Mn\textsubscript{50} one can see that faint stripes are superimposed onto small domains with magnetization pointing up or down. The faint stripes are seen better in Fig. 3.11, which shows domain images obtained at the Co L\(_3\) edge (left) and Ni L\(_3\) edge (right) at a different position of the same sample. The Fe\textsubscript{50}Mn\textsubscript{50} thickness varies here from 9.8 ML to 16.2 ML. The Co image shows domains of different gray scale corresponding to two different magnetization directions as labelled by arrows. Following one of the stripes in the Ni image from bottom to the top of the image a jump of the stripe contrast is seen. It corresponds to a change of magnetization direction of Co along the diagonal of the image. While the Co/Fe\textsubscript{50}Mn\textsubscript{50} and Ni/Fe\textsubscript{50}Mn\textsubscript{50} type of interface coupling is always the same, either FM or AFM, for a certain Fe\textsubscript{50}Mn\textsubscript{50} thickness, a change of the Co bottom layer magnetization induces a change of the upper Ni layer magnetization. The stripes are thus interpreted as a canting of the Ni magnetization direction away from pure out-of-plane directions by an oscillatory coupling to the Co layer across the AFM Fe\textsubscript{50}Mn\textsubscript{50} with 2 ML period.

From the above results, it can be concluded that the Ni layer experiences two effect from the underlying Fe\textsubscript{50}Mn\textsubscript{50}/Co bilayer: one is a spatially fluctuating pinning of the out-of-plane component of magnetization, leading to the occurrence of the small out-of-plane domains, the other is an oscillatory coupling of the in-plane component of magnetization, which leads to a periodic canting of the Ni magnetization into directions parallel and antiparallel to the Co magnetization. An in-plane spin component in Fe\textsubscript{50}Mn\textsubscript{50} which changes direction with each additional layer, pinned to the Co layer, can be responsible for the 2 ML Fe\textsubscript{50}Mn\textsubscript{50} thickness periodic canting of the Ni magnetization direction. In addition, the antiferromagnetism of the Fe\textsubscript{50}Mn\textsubscript{50} layers leads to the occurrence of small domains in the out-of-plane component of the top Ni layer.

In the following, the question of the origin of these small domains in the Ni image with stripes superimposed, was tested in another experiment. In a previous study it has been shown that deposition of 3 ML Co on 15 ML Ni induces a spin reorientation transition of the Ni magnetization from out-of-plane to in-plane [93]. Based on these previous experimental observations, the study of these small domains is addressed by depositing 3 ML Co on top of Ni/Fe\textsubscript{50}Mn\textsubscript{50}/Co trilayer. Fig. 3.12 shows domain images obtained at the Co L\(_3\) edge (left) and the Ni L\(_3\) edge (right) after deposition of 3 ML Co. The directions of crystallographic axes and incoming light are indicated at the left side of each image. The Co signal “seen” in this case is a superposition of the signal coming from the top 3 ML and from the bottom Co layer. This can be clearly recognized in the top image of the Co image map, Fig. 3.12. The stripes in the Co appear with a different contrast below and above a diagonal of the top Co image, from bottom right to the upper left. Stripes in the left bottom part show two different gray level: dark and bright. In the upper right part stripes appear heavy dark and light gray. These different contrast levels seen in the upper right and bottom left part of the top Co image must not be associated with other directions of magnetization of the top layer than along the [100] and [1\(\bar{1}\)0] axes, but appear due
Figure 3.12: Layer-resolved magnetic domain images of a 3 ML Co/15 ML Ni/3–16 ML FeMn/3–8 ML Co trilayer on Cu(001). A map of Co domain images is shown on the left hand side, the map of the Ni domain images on the right hand side. The Co and Ni magnetizations are in-plane collinear aligned. Arrows in the Co image correspond to magnetization direction of the top and bottom layers.

3.3 The Ni/Fe_{50}Mn_{50}/Co trilayer system

to the above mentioned averaging of the signal measured at the Co \( L_3 \) edge along the thickness of the layers. In the upper right part of the image, the same region as in Fig. 3.11, the magnetization of the Co bottom layer is pointing along [100] and appears with a dark contrast. In the bottom part of the Co image, the magnetization of the Co bottom layer is oriented along [100], and appears brighter. The magnetization of the top Co layer oscillates with 2 ML Fe_{50}Mn_{50} thickness, from [100] to [100], showing bright and dark contrast. The white/dark contrast of the bottom Co layer adds to the contrast at the Co \( L_3 \) edge giving rise to intermediate gray levels contrast on different sides of the image. Following one of the vertical white stripes in Co, from lower to upper side, a change of the contrast from white to heavy dark is seen. For a certain Fe_{50}Mn_{50} thickness, the interlayer coupling between the bottom and top Co layers is either FM or AFM. A change of the direction of magnetization of the Co bottom layer will induce a change
of the magnetization direction of the Co top layer, while the type of coupling is maintained.

Inspection of the contrast of the Ni image shows that the deposition of 3 ML Co induces indeed a spin-reorientation-transition of the Ni from canted out-of-plane to in-plane. A periodic oscillatory stripe domain pattern with 2 ML period as a function of Fe\textsubscript{50}Mn\textsubscript{50} thickness is observed when following the Ni image from left to the right, Fig. 3.12. Since the Co bottom layer is presumably in single domain in most of the range of the images and just changes magnetization direction at the domain boundary at the top of the images, this indicates an oscillating coupling with the Fe\textsubscript{50}Mn\textsubscript{50} thickness. The out-of-plane small domains observed in the as-grown state in Ni above 7 ML Fe\textsubscript{50}Mn\textsubscript{50} have disappeared. Analysis of the contrast reveals that the magnetization in these stripes is indeed parallel and antiparallel to the magnetization of the Co bottom layer.

Now, when following the Ni image from lower to the upper side it can be seen that the phase of coupling is periodically modulated as a function of the Co bottom layer thickness. Actually, this is similar to the step-wise character observed in the Fe\textsubscript{50}Mn\textsubscript{50} thickness region of 5 up to 12 ML in Fig. 3.10 and in Fig. 3.6.

Analysis of the Ni image, Fig. 3.12, reveals a correlation of the magnetic domain structure with the one of Co. Moreover, above 7 ML Fe\textsubscript{50}Mn\textsubscript{50}, over the entire image a periodic oscillatory coupling with 2 ML Fe\textsubscript{50}Mn\textsubscript{50} is observed in the Ni and top Co layer. A slightly different domain pattern is seen below 7 ML Fe\textsubscript{50}Mn\textsubscript{50}, the left bottom side of the each image, Co and Ni, respectively. In spite of stripes of 1 ML Fe\textsubscript{50}Mn\textsubscript{50} thickness width, domains with a less defined shape appear. Looking backward to Fig. 3.10, 7 ML Fe\textsubscript{50}Mn\textsubscript{50} is the thickness where the small domains appear in Ni, and the Fe\textsubscript{50}Mn\textsubscript{50} changes from paramagnetic to antiferromagnetic. It is the same thickness of Fe\textsubscript{50}Mn\textsubscript{50} where a transition from a 2 ML Fe\textsubscript{50}Mn\textsubscript{50} oscillatory coupling to a less ordered domain structure is seen in Fig. 3.12.

The period of all oscillations is determined with a relative error of about ±10%. Whereas ±10\% of the absolute thickness prohibits any statements about the absolute phase of the oscillations.

The periodical oscillation within the monolayer range of the Fe\textsubscript{50}Mn\textsubscript{50} thickness suggests a correlation with the antiferromagnetic structure of the Fe\textsubscript{50}Mn\textsubscript{50}. The canting of Ni within the small domains grouped in stripes is governed by the direct coupling to the Fe\textsubscript{50}Mn\textsubscript{50} layer. A non-collinear spin structure of Fe\textsubscript{50}Mn\textsubscript{50}, at least when sandwiched between Co and Ni layers, may explain the above experimental observations. This will be discussed in Sec. 5.4. The out-of-plane spin component is reflected in the presence of the out-of-plane small domains in Ni top layer, while the in-plane spin component gives rise to 2 ML Fe\textsubscript{50}Mn\textsubscript{50} thickness in-plane oscillatory coupling as evidenced by the superimposed stripes. One question that rises up in this moment is whether the non-collinear spin structure of Fe\textsubscript{50}Mn\textsubscript{50} is an intrinsic property or just a particular case, when sandwiched in-between Co and Ni layers. The in-plane spin component may be favored by the exchange coupling to the in-plane magnetization of the Co layer, while an
3.4 The Ni/Fe\textsubscript{50}Mn\textsubscript{50}/Ni trilayer system

In order to answer to this question, in the next section, the Fe\textsubscript{50}Mn\textsubscript{50} is sandwiched between two Ni layers with an out-of-plane direction of magnetization.

### 3.4 The Ni/Fe\textsubscript{50}Mn\textsubscript{50}/Ni trilayer system

In this section the coupling between two FM Ni layers across Fe\textsubscript{50}Mn\textsubscript{50} is addressed. An epitaxial Ni/Fe\textsubscript{50}Mn\textsubscript{50}/Ni trilayer has been prepared on Cu(001), in which both the Fe\textsubscript{50}Mn\textsubscript{50} and the bottom Ni layer were deposited as a crossed double wedge, rotated in the film plane by 90° with respect to each other, and the top layer was a continuous film of 15 ML Ni. As has been described in the previous section, epitaxial Ni films grown on Cu(001), above 8–10 ML, show an easy axis perpendicular to the film plane.

After deposition of the first Ni layer, an external magnetic field (∼500 Oe) in a direction perpendicular to the sample surface has been applied. This is sufficient to magnetize in saturation the bottom Ni layer, removing any internal domain structure. As discussed also in the previous section, the signal measured at the Ni L\textsubscript{3} edge contains information of both, the upper and bottom Ni layers. However, the signal attenuation by overlayers has an exponential decay as a function of the overlayer thickness. For example, in the case of Ni buried by 20 ML Fe\textsubscript{50}Mn\textsubscript{50}, the total electron yield measured at the Ni L\textsubscript{3} edge that reaches the surface is about 16% compared to that of the uncapped Ni layer [53]. In the light of the above discussion, it can be concluded that any domain structure seen in the Ni/Fe\textsubscript{50}Mn\textsubscript{50}/Ni trilayer at the Ni L\textsubscript{3} edge reflects only the magnetic structure of the upper Ni layer.

Fig. 3.13 shows the domain image obtained at the Ni L\textsubscript{3} edge. The thickness of the bottom Ni layer varies in the range of Fig. 3.13 from 0 to 15 ML, from bottom to top, as indicated at the left axis, and the Fe\textsubscript{50}Mn\textsubscript{50} thickness from 0 to 15 ML from left to right, as indicated at the bottom axis. The small arrow at the left bottom corner indicates the direction of incoming light.

Different regions as a function of the bottom Ni layer and Fe\textsubscript{50}Mn\textsubscript{50} spacer layer thicknesses can be distinguished. Now let us focus on the lower part of the image, below 8 ML Ni. Small irregular domains with an average size depending on the Fe\textsubscript{50}Mn\textsubscript{50} thickness can be observed. Between 0–3 ML Fe\textsubscript{50}Mn\textsubscript{50} a dark stripe is formed in Ni. Above 3 ML Fe\textsubscript{50}Mn\textsubscript{50} a white stripe with some very small dark domains is seen. Within 7–17 ML Fe\textsubscript{50}Mn\textsubscript{50} relatively larger domains appear in Ni. Regarded as a function of the Ni bottom layer thickness these appear elongated along the Ni wedge, with lateral dimensions decreasing with increasing Ni thickness. Within this thickness range, the Ni bottom layer is either in a paramagnetic or ferromagnetic state with magnetization oriented in-plane. Above 17 ML Fe\textsubscript{50}Mn\textsubscript{50}, in the most down-right part of the image, small domains magnetized out-of-plane are formed in Ni top layer. Since there is no Ni bottom layer in this region, the small domains are formed in the top Ni layer after deposition.
on top of 17 ML Fe$_{50}$Mn$_{50}$.

A stripe of very small domains extends from $t_{Ni} = 8$ ML, $t_{FeMn} = 5$ ML to $t_{Ni} = 5$ ML, $t_{FeMn} = 17$ ML. These small domains extend to higher Ni thicknesses only in the white domains of the stripe-like domain pattern observed for $t_{Ni} > 10$ ML. For Ni on Cu(001), 8 ML thickness corresponds to the spin-reorientation-transition from in-plane to out-of-plane direction of magnetization. Small domains are formed in the top Ni layer as a consequence of the interlayer coupling between Ni top layer with an out-of-plane magnetization direction and Ni bottom layer with an in-plane magnetization direction [93]. To lower its magnetostatic stray field energy,
domains of alternatingly up and down magnetization are energetically more favorable in the top Ni layer. Because the effective anisotropy is reduced by the coupling to the in-plane Ni magnetized layer smaller domains become energetically more favorable. Lowest the energy it cost the creation of domain walls inside which the in-plane magnetization is present. The white stripes correspond to a ferromagnetic coupling between bottom and top Ni layers. Presence of the interface roughness may introduce, in addition, an “orange-peel” like coupling which adds to the ferromagnetic coupling. Therefore, small domains are formed to minimize the energy. These small domains are not present in the black stripe because the energy of the system for an antiferromagnetic coupling is lower as compared with ferromagnetic coupling.

When the thickness of the bottom Ni layer exceeds 10 ML (the upper part of the image), oscillations of the magnetization direction as a function of Fe_{50}Mn_{50} are observed. The most left part of the image, where no Fe_{50}Mn_{50} is deposited, the white single domain reveals an out-of-plane direction of magnetization of Ni bottom layer. In the thickness range of 0 to 3 ML Fe_{50}Mn_{50}, a stripe-like domain is seen with very small features inside. In the previous section when Fe_{50}Mn_{50} is deposited on Co, below an Fe_{50}Mn_{50} thickness of about 3 ML, it becomes fully ferromagnetic by the proximity of the ferromagnetic Co layer. Assuming a similar ferromagnetic behavior of Fe_{50}Mn_{50} when deposited on Ni, a ferromagnetic coupling between the bottom and top Ni layers may be expected. A comparison of this stripe with the white single domain where no Fe_{50}Mn_{50} tells us that both Ni layer may be in-plane here. Indeed, analysis of the gray scale of Ni reveals a canting towards in-plane of the Ni magnetization direction.

When following the upper part of Fig. 3.13 an oscillatory change of the top layer magnetization direction as a function of Fe_{50}Mn_{50} thickness can be seen. From a fit of the oscillatory behavior of the Ni magnetization a long oscillation period with 5 ML period and a short oscillation period of 2 ML period with the Fe_{50}Mn_{50} thickness are deduced. The short period of oscillations are present only above 8 ML Fe_{50}Mn_{50}. As seen in the previous section, Sec. 3.3, when Fe_{50}Mn_{50} sandwiched in between two FM layers, the phase transition from paramagnetism to antiferromagnetism occurs at room temperature at about 7 ML thickness. Hence, the presence of small period of oscillation might be related with the ordered antiferromagnetic state of Fe_{50}Mn_{50}.

Studies of the growth and surface morphology of Ni on Cu(001) have reported a transition to a “multilayer” growth with well-separated 3D rectangular islands at a thickness of 4 ML [94]. Post annealing, of the deposited film, at 450 K causes a coalescence of the islands, resulting in a morphology with a smoother surface, similar to that of layer-by-layer growth, without segregation of Cu [95].

The presence of two periods of oscillation in Fig. 3.13 is addressed in a second experiment where the bottom Ni layer has been annealed in order to smooth the surface. After deposition of 15 ML Ni as a continuous film the sample was annealed for 15 minutes at 450 K. After cooling down to room temperature, an external magnetic field (≈ 500 Oe) has been applied in a
direction perpendicular to the sample surface. On top of this, 24 ML Fe\textsubscript{50}Mn\textsubscript{50} were deposited as a wedge, followed by deposition of another 15 ML Ni as a continuous film.

![Layer-resolved magnetic domain images of a 15 ML Ni/0–24 ML Fe\textsubscript{50}Mn\textsubscript{50}/15 ML Ni trilayer epitaxially grown on Cu(001).](image)

Figure 3.14: Layer-resolved magnetic domain images of a 15 ML Ni/0–24 ML Fe\textsubscript{50}Mn\textsubscript{50}/15 ML Ni trilayer epitaxially grown on Cu(001). The image represents the domain pattern of the top Ni layer. After deposition of the first Ni layer, the sample was annealed at 450 K for 15 minutes and cooled down to room temperature before deposition of Fe\textsubscript{50}Mn\textsubscript{50}. The direction of the top Ni layer magnetization shows an out-of-plane oscillatory change with a period of 2 ML Fe\textsubscript{50}Mn\textsubscript{50} thickness.

Fig. 3.14 presents the magnetic domain image obtained at the Ni $L_3$ edge. The Fe\textsubscript{50}Mn\textsubscript{50} thickness varies in the range of Fig. 3.14 from 0 to 24 ML from left to right, as indicated at the bottom axis. The direction of the incoming light is indicated by a small arrow at the upper left side of the image.

The left most part of the image in the region without Fe\textsubscript{50}Mn\textsubscript{50} and in the right most part of the image, on the Fe\textsubscript{50}Mn\textsubscript{50} plateau, both layers are in a single domain state. Below 3 ML Fe\textsubscript{50}Mn\textsubscript{50} a black stripe with small features at the edges is seen. Similar to the case of the not-annealed sample, in this thickness range the magnetization of the Ni layer is canted towards in-plane directions. The black stripe seen at about 4 ML Fe\textsubscript{50}Mn\textsubscript{50} in Fig. 3.13 is also present in the Fig. 3.14. A periodic out-of-plane oscillatory coupling with 2 ML period starts above 7 ML Fe\textsubscript{50}Mn\textsubscript{50}. For the as-grown sample, Fig. 3.13, the short period oscillations are also present only above 7 ML Fe\textsubscript{50}Mn\textsubscript{50}. The increased width of the right most dark stripe is probably just an artifact related to the rounded beginning of the plateau at the upper end of the Fe\textsubscript{50}Mn\textsubscript{50} wedge. The irregular position of the 24 ML abscise axis is intended to illustrate this effect.

In order to gain deeper insight into the spin structure of Fe\textsubscript{50}Mn\textsubscript{50} in these trilayers, the same procedure as in the previous section, Sec. 3.3, is applied. Deposition of a Co layer,
3.4 The Ni/Fe$_{50}$Mn$_{50}$/Ni trilayer system

Figure 3.15: Layer-resolved magnetic domain images of 0–2 ML Co/15 ML Ni/0–24 ML Fe$_{50}$Mn$_{50}$/15 ML Ni/Cu(001). The map of Co domain images is shown at the bottom side, the map of the Ni domain images at the upper side. Deposition of more than 0.5 ML Co top layer induces a spin-reorientation-transition in Ni from out-of-plane to in-plane. Small domains, in-plane magnetized are present only above 12 ML Fe$_{50}$Mn$_{50}$ thickness.

with an in-plane anisotropy, can induce a spin-reorientation-transition of the top Ni layer from out-of-plane to in-plane.

Fig. 3.15 shows domain images obtained at the Ni $L_3$ edge (upper panel) and (Co $L_3$ bottom panel) of the 0–2 ML Co deposited as a wedge on top of the same Ni/Fe$_{50}$Mn$_{50}$/Ni trilayer. The Fe$_{50}$Mn$_{50}$ thickness varies in the range of Fig. 3.15 from 0 to 24 ML Fe$_{50}$Mn$_{50}$ from left to right, as indicated at the bottom axis, and the Co thickness from 0 to 2 ML from bottom to top, as indicated at the left axis. The direction of the incoming light is illustrated by the arrows at the left bottom corner of each image.

The bottom part of the Co image, of Fig. 3.15, below 0.5 ML Co, shows the same oscillatory
changing of the magnetic contrast as before Co deposition, Fig. 3.14, with 2 ML Fe\textsubscript{50}Mn\textsubscript{50} period. The out-of-plane Ni anisotropy obviously turns the Co magnetization out-of-plane within this region. The deposition of the top Co layer with a thickness below 0.5 ML does not exert any influence on the Ni magnetization direction.

For a Co thickness above 0.5 ML, the direction of the Ni magnetization changes from out-of-plane to in-plane as indicated by small arrows. The wide white out-of-plane stripe seen in Ni at 5 ML Fe\textsubscript{50}Mn\textsubscript{50} for \( t_{Co} < 0.5 \) ML, forms relatively small domains, magnetized in plane, after deposition of more than 0.5 ML Co. Above 6 ML Fe\textsubscript{50}Mn\textsubscript{50} elongated in-plane domains are seen. In this region some correlation between the Ni out-of-plane stripe like magnetic domains structure before Co deposition and the in-plane domains after the spin-reorientation-transition induced by the Co layer is observed. For an Fe\textsubscript{50}Mn\textsubscript{50} thickness above 12 ML very small domains, most likely magnetized in-plane, are formed in the top FM layer, exactly as they are observed in as-grown Co/Fe\textsubscript{50}Mn\textsubscript{50} bilayers [15]. This means that in the entirely out-of-plane magnetized Ni/Fe\textsubscript{50}Mn\textsubscript{50}/Ni trilayer, there must be still a statistically fluctuating in-plane component which can force the in-plane direction of magnetization of the top FM layer to a small domain configuration. Such small domains are energetically unfavorable and disappear as soon as the antiferromagnetic layer is heated above the Néel temperature [15]. They do not occur in trilayers with identical anisotropy of both ferromagnetic layers, or in bilayers in which the ferromagnetic layer is deposited first, see for example Fig. 3.6, page 38. The presence of small domains in the top ferromagnetic layer in trilayers with different anisotropy, however, shows that perpendicular spin component behaves independently, as if the bottom ferromagnetic layer was not present. A three-dimensional spin structure in the Fe\textsubscript{50}Mn\textsubscript{50} layer has consequently to be considered. This will be discussed in Sec. 5.4.