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Preface

Experimental and theoretical research carried out at the Max Planck Institute of Microstructure Physics is primarily focused on solid state phenomena that are determined by small dimensions and surfaces and interfaces. The investigations concentrate on establishing relations between the magnetic, electronic, optical, and mechanical properties of solids and their microstructure. Thin films and surfaces are investigated as well as nanocrystalline materials, phase boundaries and defects in bulk crystals. The results of the research will provide the necessary information for creating new and improved functional or structural materials in application areas such as sensorics, opto- and microelectronics.

More than ten years ago, in 1992, the Max Planck Institute of Microstructure Physics was founded as the first institute of the Max Planck Society in the Neue Länder. The institute consists of two experimental departments (I and II) and the theory department, the research fields of which are outlined on the following three pages.

The staff of the institute, including scientific, technical and administrative personnel, comprises 99 positions, partly occupied by non-tenured personnel (19 scientists and 4 technicians). In addition, 44 co-workers have been funded by outside sources (incl. 16 graduate students) and 26 graduate students and 38 postdocs by MPG fellowships. Furthermore, 102 scientists (61 person-years) from abroad worked at the institute (incl. 26 graduate students).

Experimental Department I

We do basic research on magnetic properties of materials at reduced dimensionality. This includes magnetic surfaces, thin films, wires and dots with linear scales of 1 to 1,000 atoms. We are particularly interested in the correlation between structural properties and growth modes of these structures on the one hand, and their magnetic, electronic, and mechanical properties on the other hand.

Thin films are grown by molecular beam epitaxy and/or laser ablation. Magnetic wires and dots are made by using specially structured substrates for molecular beam epitaxy. Surface structures are analyzed by X-ray diffraction, scanning tunneling microscopy, scanning Auger microscopy, low energy electron diffraction, and photoemission microscopy. Magnetic properties of surfaces and interfaces are analyzed by spin-polarized low energy electron scattering, spin-polarized scanning tunneling microscopy and magnetic second harmonic generation by fsec lasers. The electronic structure of films and alloys is studied by spin-polarized photoemission and magnetic dichroism using synchrotron radiation. Using photoemission microscopy, magnetic microstructures can be analyzed in an element-specific way. The magneto-optic Kerr effect serves as a transfer standard between different experiments. Our present interests also include oxides, either as ferromagnets in their own right or as insulating films in spin-polarized tunneling devices for magnetoelectronics applications.

Since March 2003 we have an intense and fruitful cooperation with the Laboratoire Louis Néel, Grenoble, within the framework of a "Joint European Laboratory" (Laboratoire Européen Associé). This includes regular meetings at Grenoble or Halle as well as exchange of personnel lasting from a few weeks to one year.

Experimental Department II

Modern society demands better and more sophisticated materials for information, communication, engineering as well as bio-technological applications. The desirable properties of these materials are frequently based on a detailed control of their nanostructure. Our basic research is aimed at supplying the scientific understanding for the design and fabrication of such improved or completely new materials, thus pushing the limits of nanoscience and nanotechnology.

Our capabilities include methods to fabricate nanocrystallites, nanowires and nanotubes, micro- and macroporous silicon and nanoporous aluminum oxide as well as nanoporous polymers. Molecular beam epitaxy, wafer bonding, laser deposition as well as clean room facilities are available. Advanced electron optical imaging techniques are indispensable research tools. They are accompanied by image simulations and/or molecular dynamics simulations.

The materials dealt with include a variety of different semiconductors as well as ferroelectrics, polymers, glasses, ceramics and composites. In the semiconductor field we focus on the study of quantum dot structures involving III-V compounds or silicon and germanium. We recently launched activities on flexible semiconductor materials including nanowires in collaboration with the Physics Department of the Martin Luther University and the Fraunhofer Institute of Mechanics of Materials at Halle.

Silicon photonics has developed into one of the main research areas of the department. Experimental and theoretical research activities include high quality two-dimensional photonic crystals, layers of silicon nanocrystallites of controlled size and indium arsenide or germanium inclusions in silicon for potential silicon light emitters as well as silicon nanowires. Silicon membranes with ratched-like modulated pore profiles are applicable as tunable pumps for mesoscopic particles. Wetting of porous templates allows to form arrays of nanotubes from many materials. Moreover, various nano-biotechnology projects are based on ordered pore systems. Nano-imprint technology has been at the center of activities for creating long-range ordered pore arrays in aluminium oxide. In the area of “Multifunctional Nanorods” a federally funded junior research group has started its activities end of 2003.

Further investigations on the fabrication and properties of ferroelectric nanostructures and single crystalline ferroelectric layers of high polarisation were carried out. Nano-imprint technology has been established to fabricate such structures.

One of the mature research areas of the department is that of wafer bonding. Wafer bonding allows a new design freedom for materials combinations in a wide array of areas from microelectronics and microsystems to photonics. Recently, wafer bonding has internationally gained renewed interest as the preferred approach to fabricate biaxially and uniaxially strained silicon layers on insulating substrates, an area we are now also actively pursuing in close collaboration with industrial partners.

Theory Department

The Theory Department has started its activities in April 1998 and carries out theoretical research on the electronic, magnetic, optical and electrical properties of micro- and nanostructured solid state systems. More specifically, our research is currently focused on (i) exchange interactions in magnetic ultrathin films and nanostructures, (ii) magneto-electronics, i.e., spin-dependent electronic transport phenomena, and (iii) electron correlation spectroscopies of solid surfaces.

The exchange interaction is the "glue" which maintains the spins oriented parallel to each other in a ferromagnet. It is thus of primary importance for determining the Curie temperature of ferromagnetic systems. It is rather well understood in bulk systems, but not in low-dimensional systems such as ultrathin films and nanostructures. Our investigations are based upon first-principles electronic structure calculations. In particular we are investigating ordered and disordered transition methods, low-dimensional systems such as ultrathin films and nanoclusters, and novel materials such as dilute magnetic semiconductors. These intensive numerical calculations are complemented by simple model studies, which provide the basis for the physical interpretation of the results.

Magneto-electronics is an emerging field of science and technology in which the spin-dependent electronic transport properties of micro- and nanostructured solid state systems are studied and used to create new devices. Our research in this field is concerned with the tunneling magneto-resistance in ferromagnet/insulator/ferromagnet systems, with spin-dependent transport phenomena in a two-dimensional electron gas such as the Rashba effect with the anomalous Hall effect of ferromagnets, and with magnetic scanning tunneling microscopy. Our approach relies on analytical and numerical calculations on simplified models from which we obtain some insight on the physical mechanisms involved, combined with sophisticated first-principle calculations.

Electron coincidence spectroscopies such as (e,2e) spectroscopy and double photoemission have been recently successfully performed on solid state surfaces in our Institute. In contrast to single-electron spectroscopies, these new techniques yield some information on electronic pair correlations in the solid. We are developing theoretical and computational methods in order to investigate the physical mechanism of double emission processes and to interpret the electron coincidence experiments. These methods are applied to various nanostructured systems such as fullerenes, nanowires etc.

Selected Results

High wave vector spin waves in ultra-thin hcp Co-films

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Magnetic excitations determine fundamental properties of magnetic materials. Collective magnetic excitations, so called spin waves, are important because they are, for example, responsible for the drop of the magnetization of a magnet with increasing temperature. By the study of spin waves, one gains insights into the magnetic interactions in solids. For different wavelength of spin waves, different interactions define the excitation energies. For wavelengths of the order of the atomic distances, that is for high wave vector spin waves, the excitation energy is solely determined by the microscopic exchange interaction. Spin polarized electron energy loss spectroscopy (SPEELS) is the only technique that allows the study of these high wave vector spin waves in ultra-thin films [1, 2]. Here, we report new results obtained by SPEELS measurements on 8 atomic mono-layer (ML) thick, hexagonal closed packed (hcp) Co-films grown on W(110). In this study, we could measure spin-wave excitations with wave-vector transfers up to 1.64 Å^{-1} which corresponds to the surface Brillouin zone boundary ($\bar{\text{K}}$-point) of the system. At this point, the wavelength of the spin wave equals the length of the periodicity of the underlying crystalline structure.

In our experiments, we used a spectrometer especially developed to investigate spin-wave excitations. A photo of this spectrometer is shown in Fig. 1. In the experiments, the spin-polarized electron beam is created by a strained GaAs-photocathode. Before the electron beam hits the sample, it is monochromatized by electrostatic monochromators. The scattered electron beam is analyzed with respect to the energy and wave vector which the electrons transferred to the sample. A detailed description of the SPEEL-spectrometer can be found in Ref. [3].

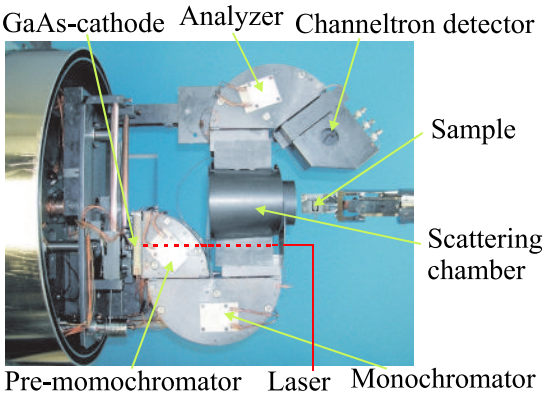

Fig. 1: Photo of the SPEEL-spectrometer used in the experiments. As the source of spin-polarized electrons, we use a strained GaAs-photocathode. The emitted electron beam is monochromatized before scattering from the sample. The scattered electrons are analyzed with respect to their wave vector and energy transfer.

The quantity measured in the experiments is the intensity of the scattered electron beam. This is done as a function of the energy transfer for the two possible orientations of the incoming electron spin with respect to the sample magnetization and for a certain wave-vector transfer. An example of the measured energy loss spectra is shown in Fig. 2. A pronounced loss feature caused by spin-wave excitations is visible in the $I_{\downarrow}$-spectrum (incoming electron spin is parallel to the spin of a minority electron in the sample). The conservation of the total angular momentum during the scattering process forbids the excitation of spin waves for incoming electrons having a majority spin character ($I_{\uparrow}$). This spin selective excitation process allows to distinguish a spin-wave loss feature from other excitations. For example, vibrations of minor amounts of adsorbates create small loss features in the spectra. The excitation probability of these vibrational losses is (almost) independent of the electron spin. Therefore, these loss features are visible in

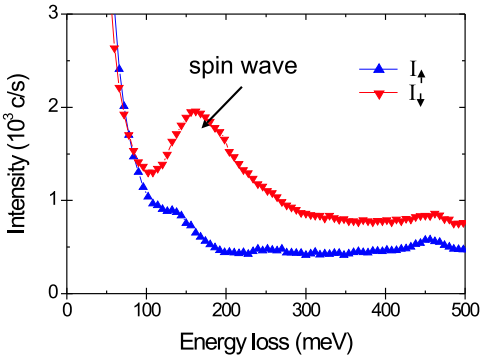

Fig. 2: Two SPEEL-spectra measured on 8 ML hcp Co on W(110) for a wave-vector transfer of 0.78 $Å^{-1}$. $I_\uparrow$ ($I_\downarrow$) denotes the spectra recorded for the incoming electron spin direction parallel (antiparallel) to the spin of a majority electron in the sample. In the $I_\downarrow$-spectrum a prominent spin-wave loss feature is visible at about 170 meV energy loss. The presented spectra have been corrected for the incomplete polarization of the incoming electron beam. The energy resolution in this scan was about ΔE = 40 meV (FWHM), the kinetic energy of the incoming electrons was E_{kin} = 4 eV.

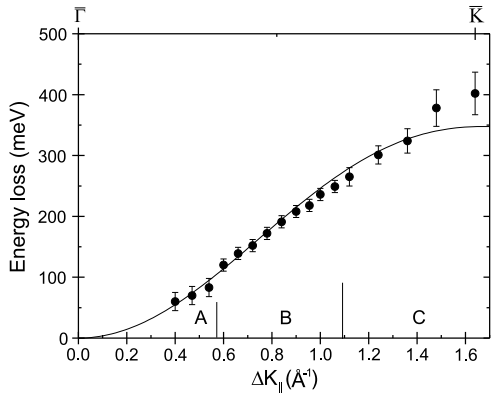

Fig. 3: Spin-wave dispersion of 8 ML hcp Co on W(110) measured by SPEELS. The points indicate the center of the spin-wave peaks in each spectra. $\bar{\Gamma}$ and $\bar{K}$ mark the origin and the boundary of the surface Brillouin zone of the system. The three different regions A, B and C have been measured with different spectrometer settings. In region A, the spectra were recorded with higher energy resolution $\Delta E \approx$ 20 meV, while for B and C $\Delta E \approx$ 40 meV. For region A and B E_{kin} = 4 eV. For C E_{kin} = 25 eV.

both spectra shown in Fig. 2 (vibration are caused by H (130 meV), CO (240 meV) and H_20 (450 meV)).

Spin waves follow a characteristic dispersion which links their wave vector with their energy. Several spectra have been recorded for different wave-vector transfers to investigate the spin-wave dispersion of the system. The resulting dispersion curve is presented in Fig. 3. The dispersion is in surprisingly good agreement to the dispersion of a surface spin wave derived from a nearest neighbor Heisenberg model. The fit of this dispersion to the experimental data is shown in Fig. 3 as a solid line. The only parameter used to fit the calculated dispersion to the data is the product of the exchange coupling constant and the magnetic moment JS. From the fit, a value of JS =14.8$\pm$1 meV results.

Spin waves in bulk hcp Co have been measured by inelastic neutron scattering for wave-vector transfers up to about 0.8 $Å^{-1}$ [4]. The spin-wave energies obtained by neutron scattering on bulk Co are slightly higher than the spin-wave energies measured by SPEELS on an 8 ML Co-film. Within the nearest neighbor Heisenberg model, the differences cannot be explained by the differences between the surface and bulk modes.

The dispersion measured by SPEELS is in reasonable agreement to theoretical predictions, see for example Ref. [5].

In conclusion, we have investigated spin waves at surfaces of 8 ML hcp Co on W(110) by SPEELS. We were able to measure the dispersion curve up to the surface Brillouin zone boundary.

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Imaging the exchange correlation hole in a (γ,2e) experiment from NaCl

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Correlation between electrons is a fundamental characteristics of systems containing many electrons. A quite promising way to investigate correlation effects is to study the double photoemission (DPE), where the absorption of a single photon leads to the simultaneous excitation and emission of two (correlated) electrons. This simultaneous two-orbital excitation following the absorption of one VUV photon is exclusively caused by the interelectronic interaction because in the case of exitations by electron dipole transitions, a photon can only interact with one electron at a time[1],[2]. Recent theoretical calculations, carried out by Fominykh et al[3] on Cu(001) at a photon energy of $h\nu = 21.2$ eV, yield a striking structure in the angular distribution of photoemission from a surface, namely zones of strongly reduced intensity when distribution of the momenta parallel to the surface of one electron with a certain energy is mapped while the second electron is fixed in its momentum at the same energy. This observation can be understood from electron-electron repulsion, where the exchange interaction prevents two electrons from escaping with comparable wave vectors within a proximity determined by the screening lenght. Therefore, this so-called *exchange correlation hole* can be seen as a direct manifestation of exchange and correlation between two photoelectrons, where the size of the hole reflects the strength of the interaction between two electrons.

Energy- and angle-distributions of correlated photoelectrons ejected from a NaCl(100) single crystal surface were studied in order to image the *exchange correlation hole* experimentally. In this experiment, the time-of-flight technique was utilised to determine the energy of the electrons released from the surface by an incoming photon.

The momentum of one electron is determined by using a microchannel plate (MCP) in combination with a position sensitive resistive anode, and the corresponding coincident electron is detected with a central collector. The surface normal of the probe was pointing to the center of the detector, marking the $0°$ polar angle of the emission direction. A schematic view of the experimental set up is given in fig. 1.

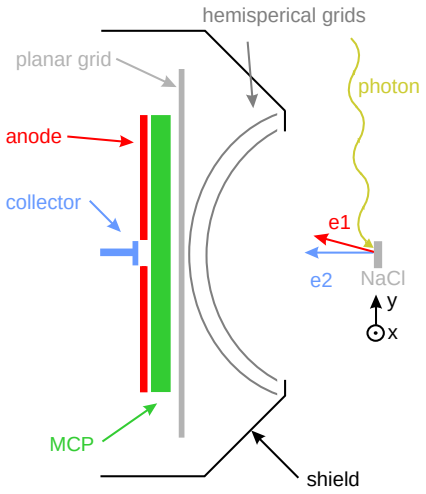

Fig. 1: Schematic view of the experimental set up. e1, e2: photoelectrons, MCP: multichannelplate.

Data were taken at a photon energy of $h\nu = 34$ eV with non-polarised synchrotron light at *BESSY II*.

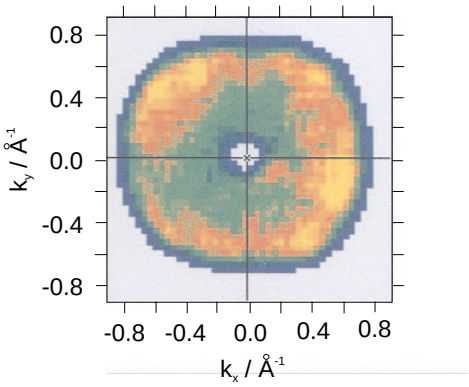

Fig. 2: Momentum distribution of one electron while the momentum of the second electron is fixed around the surface normal. The energies of the two electrons are 8 eV and 7 eV, respectively.

Fig. 2 shows the momentum distribution of the first photoelectron with a kinetic energy of 8 eV, if its measured intensities of the momenta

parallel to surface k_x and k_y are plotted, while the momentum of the second (correlated) electron with an energy of 7 eV is fixed along the surface normal. In fig. 2, the position of this second electron, which corresponds to the *blue* electron e2 in fig. 1, can be seen as a white spot in the center. The most striking feature in fig. 2 is an area of rather low intensity that appears almost as a disk with a diameter of about 1 Å^{-1} around the center. This structure is principally very much alike to the calculated angular disitribution of photoemission performed by Fominykh et al[3], and implies that the lack of intensity reflects the *exchange correlation hole*, as dicussed above.

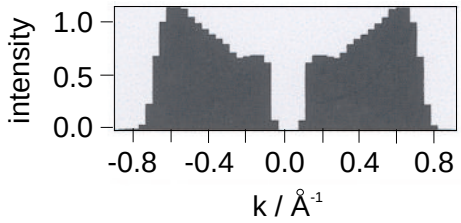

Fig. 3: Symmetrised radial line scan of the momentum distribution shown in fig. 2.

It should be noted that the full momentum distibution cannot be deduced from this data, since the accessible momentum range is stricly limited by the finite size of the MCP, so that the apparent structure in fig. 2 at the edge of the spectrum is meaningless. The actual dimension of the correlation hole becomes clearer from the symmetrised radial line scan done to the data in fig. 2, where the measured intensity is summed up along the parallel momentum k, see fig. 3. Here it is interesting to note that the dimensions of the calculated copper case and measured NaCl are comparable, which suggests that the size of the hole is not very much depending on the photon energy and the probe material.

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Surface alloy formation and stress oscillations of FeMn monolayers on Cu(001)

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Magnetic multilayers offer unique properties in applications as sensors. These sensors exploit the dependence of the electrical resistance of a multilayer stack of ferromagnetic (e.g. Fe) and paramagnetic (e.g. Cu) materials on the relative orientation of the magnetization direction in the different layers: the resistance is low for a parallel orientation of the magnetization directions, and it is high for antiparallel orientation. Highly sensitive magnetic sensors can be constructed if one succeeds to obtain one ferromagnetic layer with a magnetization direction, which follows easily an external field, whereas the magnetization direction of another magnetic layers is pinned along a certain direction. Sensors of this kind are found in read heads of modern computer hard disks, where the faint magnetic stray field of the magnetically stored information is detected as a change of resistance.

An essential aspect of these magnetic sensors is the pinning of the magnetization direction of a ferromagnetic layer. This is achieved by growing a ferromagnetic layer on top of an antiferromagnetic layer (e.g. FeMn). The so-called exchange interaction between these layers pins the magnetization direction of the ferromagnetic layer along one direction. The physics of the exchange interaction and the study of antiferromagnetic materials are topics of high current interest, due to applications as magnetic sensors, and due to interesting fundamental properties of antiferromagnetic layers, like the noncollinear spin structure of FeMn[1].

We present experimental evidence which indicates that the interface region between FeMn and Cu is structurally much more complex as one might have expected [2]. Naively, a sharp interface between the two materials might have been anticipated. However, highly sensitive stress measurements during FeMn growth on Cu(001) indicate substantial intermixing at the interface. Our measurements, in which structural aspects were clarified by surface X-ray diffraction [3], indicate that Mn from the deposited FeMn alloy replaces Cu surface atoms, and a MnCu surface alloy is formed. This renders the deposited FeMn material Fe-rich, and a modified antiferromagnetic coupling of the growing FeMn film is expected.

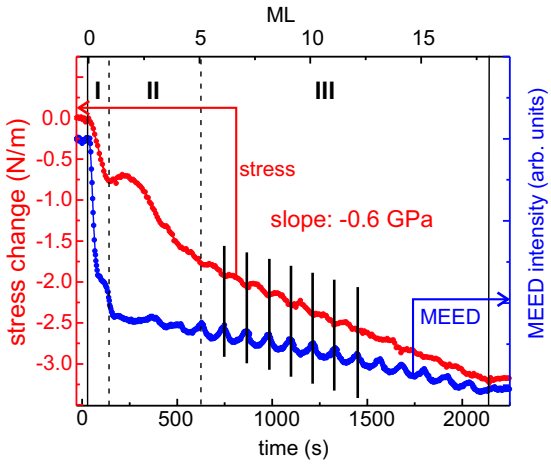

Fig. 1: Combined stress measurements (red curve, left scale) and MEED intensity measurements (blue curve, right scale), measured simultaneously during the deposition of 18 layer FeMn on Cu(001) at 300 K. Three growth regimes I–III are identified. Region I is ascribed to the formation of the MnCu surface alloy. The vertical bars in region III reveal coherent layer filling and stress variation. 1 ML: 1 atomic layer.

Stress measurements were performed by the crystal curvature technique. In short, a stress imbalance between the FeMn-covered front surface and the uncovered back surface of a 0.1 mm thin Cu(001) substrate leads to a curvature of the Cu crystal. This stress-induced curvature is measured by reflecting a laser

beam form the sample surface onto a position sensitive detector [2]. The surface roughness was monitored in situ during growth by medium-energy-electron-diffraction (MEED). Maxima in the MEED intensity are ascribed to smooth surfaces, which are indicative for a filled layer. Consequently, regularly spaced MEED intensity maxima indicate completion of successively filled FeMn layers.

Stress measurements are sensitive tools to detect structural and morphological changes of ultrathin films with sub-monolayer sensitivity. In general, film and substrate have different lattice constants, which give rise to a epitaxial misfit between both materials. This misfit strain induces a film stress. In our example, FeMn is under slight compression of -0.4 %, FeMn prefers a larger lattice constant as Cu, and a compressive stress of -0.66 GPa is expected. Indeed, our stress measurements identify this magnitude of stress in FeMn films thicker than 5 ML, see Fig. 1. The film morphology also influences the resulting stress. A film with a rough surface, which is covered by many nanometer scale islands, induces less stress as compared to a smooth, atomically flat film. Stress relaxation is effective for small islands [4], and in the initial phase of layer filling, where small islands grow in size, a reduced stress change is expected. Indeed, this correlation between layer filling by islands growth and stress relaxation is found in our combined stress and MEED measurements presented in Fig. 1 in regime III, where the vertical bars identify a coherent variation of both layer filling and stress. The largest compressive stress occurs for filled layers (MEED maxima).

Only for a film thickness in excess of 5 ML our stress and MEED measurements support the simplistic model of layer-by-layer growth of epitaxially strained FeMn. The initial strong compressive stress of -0.7 N/m after deposition of 1 ML in regime I comes as a surprise. The initial compressive stress identifies a surface alloy formation between Mn and Cu. This conclusion is supported by the individual stress measurements for Mn deposition on Cu(001) shown in Fig. 2, and by additional structure investigations by electron and X-ray diffraction [3]. The formation of the c-(2×2) Mn-Cu surface alloy induces compressive stress of -1.2 N/m at 0.5 ML, whereas the same deposition of Fe on Cu(001) does not induce any appreciable stress change. The Mn-induced compressive stress upon surface alloy formation is ascribed to the relaxation of the tensile surface stress of Cu(001) upon incorporation of the larger Mn atom [3]. The formation of the MnCu surface alloy leads to a structurally and chemically inhomogeneous interface. Regime II in Fig. 1 marks the transition towards FeMn layer-by-layer growth which occurs with increasing layer thickness.

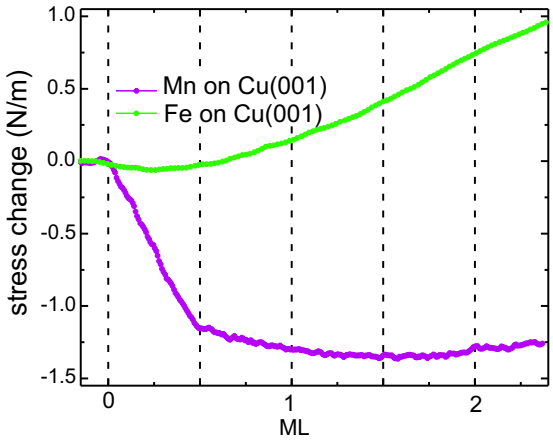

Fig. 2: Stress as a function of thickness for Fe and Mn on Cu(001) at 300 K. 0.5 ML Mn induce a compressive stress of -1.2 N/m, whereas 0.5 ML Fe do not induce an appreciable stress change [3].

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Probing the antiferromagnetic spin structure of ultrathin FeMn films

W. Kuch, L. I. Chelaru, F. Offi, J. Wang, M. Kotsugi

In ferromagnetic materials the spins of the individual atoms are subject to long range parallel order, giving rise to a macroscopically observable magnetization. In antiferromagnetic materials, in contrast, the atomic moments—although also long range ordered—average to zero within a few atomic distances, and are thus much more difficult to detect. The spin structure of many antiferromagnetic materials has been investigated in bulk samples already decades ago by methods like neutron scattering or Mössbauer spectroscopy. The spin structure of ultrathin antiferromagnetic films with thicknesses of only a few nanometers, however, has mainly remained concealed, because these methods require thick samples to obtain sufficient signal. Ultrathin antiferromagnetic films are frequently employed in magnetic thin film devices like magnetoresistive sensors to manipulate the magnetic properties of adjacent ferromagnetic layers. The knowledge of the spin structure of the antiferromagnetic layer is crucial for both the design and operation of these devices as well as the fundamental understanding of the magnetic interaction between ferromagnetic and antiferromagnetic layers.

We have used photoelectron emission microscopy (PEEM) with X-ray magnetic circular dichroism (XMCD) as magnetic contrast mechanism to study the magnetic interface coupling in single-crystalline stacks containing ultrathin antiferromagnetic FeMn layers that are sandwiched between two ferromagnetic layers. By assembling trilayer stacks that have non-collinear axes of magnetization of the ferromagnetic layers we can probe the collinearity of the antiferromagnetic spin structure.

Single-crystalline, fcc FeMn films were obtained by deposition on a Cu(001) single crystal substrate [1]. Scanning tunneling microscopy measurements had revealed that the FeMn films grow on Cu(001) in a near-perfect layer by layer-mode [2]. Some of the layers were prepared as wedges in order to explore the thickness dependence of the magnetic domain patterns. The XMCD-PEEM measurements were performed using synchrotron radiation at the BESSY synchrotron radiation source in Berlin.

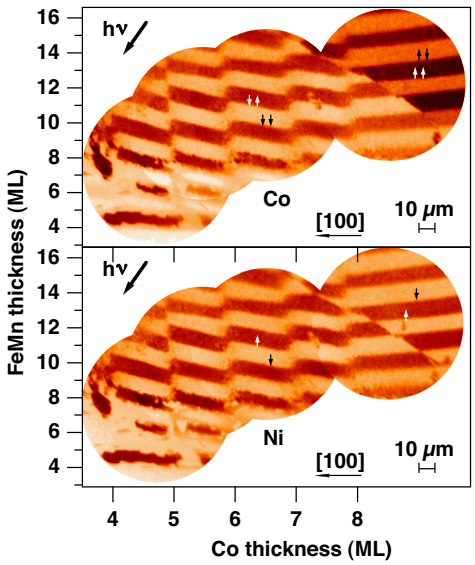

Fig. 1: Element-resolved magnetic domain images of 3 ML Co/15 ML Ni/FeMn (wedge)/Co (wedge)/Cu(001). Top: domain image at the Co L_3 edge, bottom: domain image at the Ni L_3 edge.

Fig. 1 shows element-resolved magnetic domain images of a sample in which an FeMn wedge (left axis) was sandwiched between a Co wedge as the bottom ferromagnetic layer (bottom axis), and a top ferromagnetic layer consisting of 15 atomic monolayers (ML) of Ni and 3 ML Co. The top panel shows the magnetic domain image obtained at the Co L_3 absorption edge; its signal is therefore composed by contributions of both the bottom and top ferromagnetic layers, since both of them contain Co. The bottom panel shows the magnetic domain pattern obtained at the Ni L_3 edge, which represents the domain pattern of the top ferromagnetic layer only. Regular stripes with alternating parallel and antiparallel coupling between the two ferromag-

netic layers across the antiferromagnetic FeMn spacer layer are recognized as the FeMn thickness is increased. The period of 2 ML in FeMn thickness indicates direct exchange coupling through the antiferromagnetic spacer layer. In addition to these stripes, a sawtooth-like wiggling of the phase of this periodic interlayer coupling is observed as the Co bottom layer thickness is increased. It is attributed to the periodic modulation of the bottom interface roughness due to the layer-by-layer growth of the films.

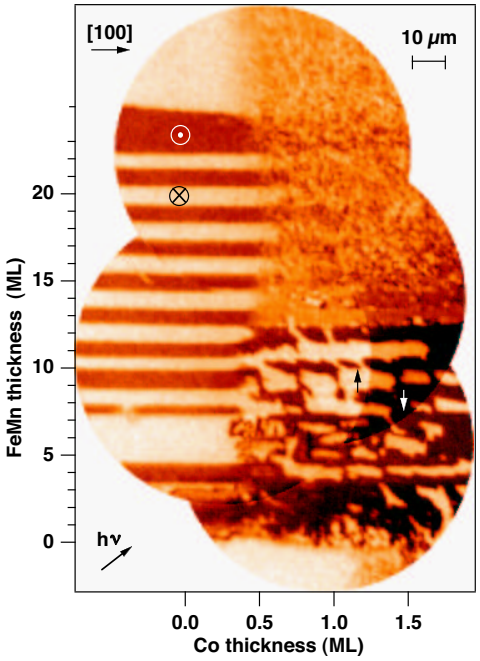

Fig. 2: Magnetic domain image at the Ni L_3 edge of Co (wedge)/15 ML Ni/FeMn (wedge)/15 ML Ni/Cu(001). In the left part of the image the magnetization of the top layer is perpendicular to the film plane, while it is in-plane for Co thicknesses above 0.5 ML.

Fig. 2 shows the magnetic domain pattern at the Ni L_3 edge of a trilayer in which an FeMn wedge (left axis) is sandwiched between 15 ML Ni on the bottom and Co (wedge, bottom axis)/15 ML Ni on the top. The bottom layer had been magnetized into a single domain state after deposition. The image thus represents the domain pattern of the top Co/Ni layer. In the left part of the image, for zero or low Co thickness, a stripe-like domain pattern with a stripe period of 2 ML FeMn thickness as before is observed at FeMn thicknesses above 9 ML. Here the magnetization directions of the bottom and top ferromagnetic layers are perpendicular to the film plane, with alternanting parallel and antiparallel alignment. At Co thicknesses above 0.5 ML the magnetization of the top ferromagnetic layer turns into the film plane. The small domains observed in this region for FeMn thicknesses above 12 ML show that in addition to an oscillatory coupling of the spin component perpendicular to the film plane, the spin component in the plane experiences a laterally fluctuating pinning [3]. This provides direct experimental evidence that a three-dimensional non-collinear antiferromagnetic spin structure must be present in the FeMn layer. Such a spin structure, the so-called $3Q$ structure, is schematically depicted in Fig. 3. It is characterized by four different sublattices in which the spins point along four different $\langle 111 \rangle$ directions.

This interesting result has also major implications on theoretical models of the antiferromagnet–ferromagnet interaction that try to describe the exchange bias effect.

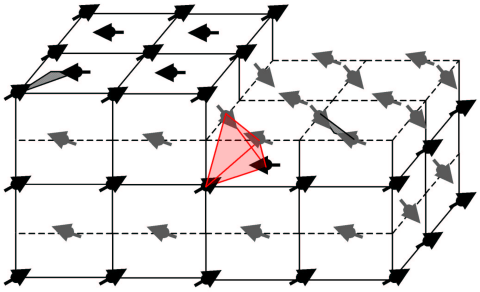

Fig. 3: Sketch of the $3Q$ non-collinear antiferromagnetic spin structure. Monatomic steps at the (001) surface exhibit 90° different axes of the spin component in the film plane, and opposite sign of the spin component perpendicular to the film plane.

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Single-crystalline Fe/Cr/Fe/MgO/Fe magneto-tunnel junctions grown on GaAs(001)

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In the last years, the spin of the electron has been considered for the design of new electronic devices. The spin degree of freedom can be used to add the capabilities of non-volatility and increased data processing speed to these devices. In particular, the injection of spin-polarized carriers at a ferromagnet-semiconductor interface is recognized as an important problem. It was successfully proven recently that a magnetic-semiconductor based system is feasible as the spin-electronic analogue of the electro-optic light modulator, however only under conditions of low temperatures and high magnetic fields. From the practical requirement of functionality with small magnetic fields and at room temperature, the concept of spin-injection from a metallic ferromagnet into a semiconductor is still very attractive. Due to a recent work that imposed severe restrictions on the functionality of integrated ferromagnet-semiconductor devices, this concept needs substantial modifications, e.g. by the integration of a tunnel junction to produce hot electrons as a part of the spintronic device [1]. Single-crystalline ferromagnet/insulator/ferromagnet magneto-tunnel structures seem to be an ideal candidate to be integrated with a semiconductor element. This is due to theoretical calculations that predict a 1000% Tunneling Magneto-Resistance (TMR) effect in the case of an ideal Fe/MgO/Fe junction.

Our single-crystalline Fe/MgO/Fe tunneling structures are grown on GaAs(001). An independent magnetization switching in the Fe films is achieved by pinning the magnetization of one of the Fe films (called ”middle”) by antiferromagnetic coupling across a Cr-spacer to another Fe film (called ”bottom”). Antiferromagnetically coupled Fe/Cr/Fe trilayers, epitaxially grown on GaAs(001), reflect magnetic anisotropy of the GaAs(001)/Fe system characterized by uniaxial anisotropy of the easy axis of magnetization along the [110] direction. The total structure of our samples is: GaAs(001)/Fe/Cr/Fe/MgO/Fe (Fig.1). The magnetization reversal in the GaAs/Fe/Cr/Fe structures can be controlled by changes of the thickness of the Fe layers. At low fields, the Fe films in the Fe/Cr/Fe trilayer switch simultaneously and keep their antiferromagnetic coupling. The switching field depends on the thickness relation between both Fe layers, i.e. the thickness of one of the layers with respect to the thickness of the complete structure [2]. The lowest switching field obtained

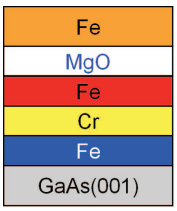

Fig. 1: Schematic view of the complete sample. The magnetization of the middle Fe layer is antiferromagnetically coupled to the bottom Fe layer. The magnetization of the top ”free-Fe” layer switches independently.

in trilayers with strongly different Fe thickness corresponds to the coercivity of the thicker layer and the highest is approached when the thickness of both Fe layers becomes similar. Depending on the magnetic techniques used for characterization, the observed magnetic properties of multilayers may vary. In our case of antiferromagntically coupled Fe layers (GaAs(001)/50MLFe/9MLCr/37MLFe), ”reversed” minor hysteresis loops are observed by the longitudinal magnetooptical Kerr effect (MOKE) (Fig.2). This means that a negative remanence is detected when the thicker bottom Fe layer is saturated along the applied field. This behavior is interpreted by depth variation of the MOKE sensitivity, which results in a smaller contribution of the thicker but deeper layer, than of the thinner middle layer, to the

total MOKE signal. The thinner middle Fe film which is magnetized opposite to the field thus contributes with a negative ellipticity and a reversed minor loop is observed. This scenario is verified by another magnetic transition which occurs even within the field range that can be applied in situ in our experimental setup (+/- 30mT). The ellipticity becomes positive due to the spin-flop transition that eliminates the negative contribution of the middle Fe layer to the total MOKE signal (Fig.2) [3]. The shape of the MOKE loops is more complex after the structure is completed by further deposition of 15ML of MgO and 15ML of Fe on top of the GaAs(001)/Fe/Cr/Fe stack. This is shown in Fig.3a. Qualitatively, the shape of the loop can be understood by combining the loop measured for the GaAs(001)/Fe/Cr/Fe sample (Fig.2) with the loop of a single uncoupled Fe film characterized by a small coercivity. Finally, the sample was coated with a protective Au film and treated for ex-situ electrical transport measurements. The junction area was reduced to $5 \times 5\ \mu\mathrm{m}^2$. Fig.3b shows the result of magneto-transport measurements. After a careful analysis of the relative magnetization orientation of the Fe electrodes in the Fe/MgO/Fe junction, it becomes obvious that the resistance changes follow exactly the changes in magnetization orientation of the electrodes. For parallel magnetization orientation (depicted by arrows colored correspondingly to the Fe layers shown in Fig.1) the junction tunneling resistance is smaller than for the antiparallel configuration. The obtained value of TMR of 10-12% is much lower than that predicted theoretically. In the case of magneto-tunnel junctions grown on GaAs substrate, it is difficult to obtain a good crystallographic order and an atomically flat surface of Fe. This is due to the limited temperature that can be applied during the growth providing no interdiffusion in the GaAs/Fe interface. Consequently, the growth of the MgO layer results in some structural imperfections reducing the TMR effect.

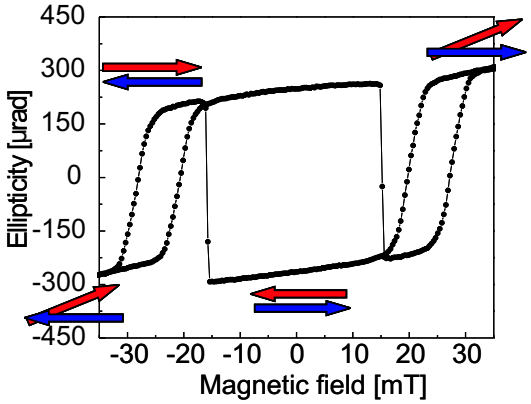

Fig. 2: MOKE loop measured at room temperature along [110] for GaAs(001)/50MLFe/ 9MLCr/37MLFe

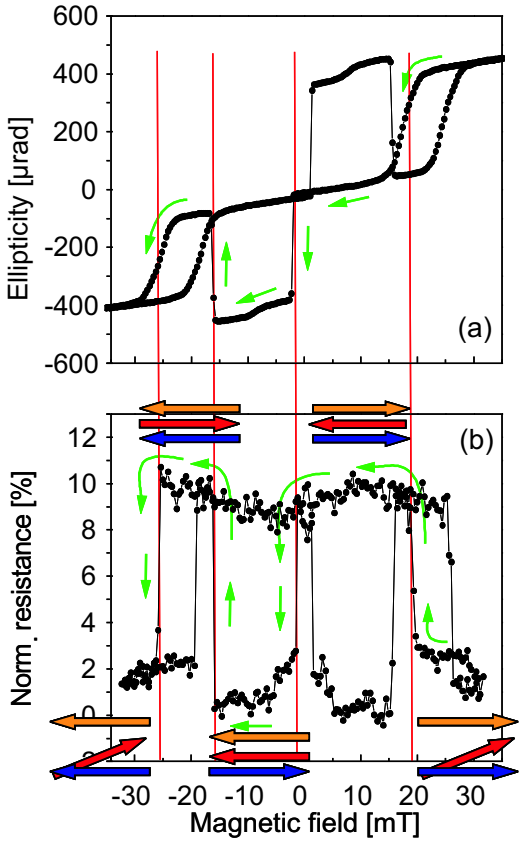

Fig. 3: MOKE loop (a) and normalized resistivity (b) measured at room temperature along [110] for 50MLFe/9MLCr/37MLFe/15MLMgO/15MLFe grown on GaAs(001)

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Exploring magnetization reversal dynamics in magnetic multilayers with temporal, spatial and layer resolution

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How long does it take to load this text from the hard disk of your computer? With the current technology this depends on the time needed to turn the magnetization of a soft magnetic layer in the read head of the hard disk drive. A magnetoresistive spin valve read head sensor, as it is nowadays commonly used in magnetic hard disk read heads, consists of two ultrathin ferromagnetic layers that are separated by a non-magnetic spacer layer. The magnetization direction of one of the magnegic layers is switched by the magnetic bits on the disk over which the head passes, leading to a change in electrical resistance. Presently, read and write times approach one nanosecond, corresponding to 1 GHz frequency. Controlling and understanding the magnetization reversal dynamics in magnetic thin films is thus a major issue for accelerating the speed at which data can be stored and read back.

Only very few experimental techniques can address the microscopic magnetization reversal behavior of the different magnetic layers in a spin valve separately. One of them is photoelectron emission microscopy (PEEM) with synchrotron radiation, which employs x-ray magnetic circular dichroism (XMCD) as magnetic contrast mechanism [1]. Because of the element selectivity of XMCD, different magnetic layers in a multilayered stack can be imaged separately. Time resolution is obtained using a pump–probe approach (Fig. 1). Images are acquired for different times before, during, or after the application of the magnetic field pulses. In this way the magnetization dynamics of each magnetic layer can be visualized separately with a time resolution limited by the X-ray pulse width (about 60 ps).

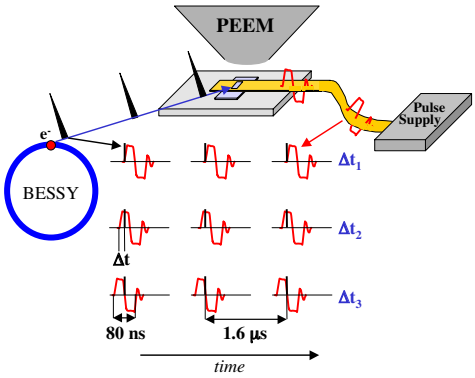

Fig. 1: Sketch of the stroboscopic time-resolved magnetic imaging technique. Current pulses from a fast pulse power supply induce magnetic field pulses of several nanoseconds (red curves) in a microcoil close to the sample position. Synchronized X-ray pulses from the synchrotron radiation light source BESSY in Berlin of about 60 ps length hit the sample at a variable time delay Δt relative to the magnetic pulses [2]. Electrons excited by these x-ray pulses are used to obtain dynamic stroboscopic and layer-resolved domain images in the PEEM.

We have used the combined temporal, spatial, and layer resolution of time-resolved XMCD-PEEM to study the magnetization reversal dynamics of a 5 nm $Fe_{20}Ni_{80}$/4 nm Cu/5 nm Co spin valve on the nanosecond time scale. Fig. 2 shows stroboscopic domain images of the $Fe_{20}Ni_{80}$ layer (domain contrast green/blue) and of the Co layer (domain contrast yellow/red). The images were acquired at different times during the application of short field pulses as indicated in the upper left panel. The magnetization of the magnetically harder Co layer is not affected by these pulses, i.e., the Co layer shows the same pattern as in panel e for all times. In the $Fe_{20}Ni_{80}$ layer, the magnetic field pulses favor the growth of the blue

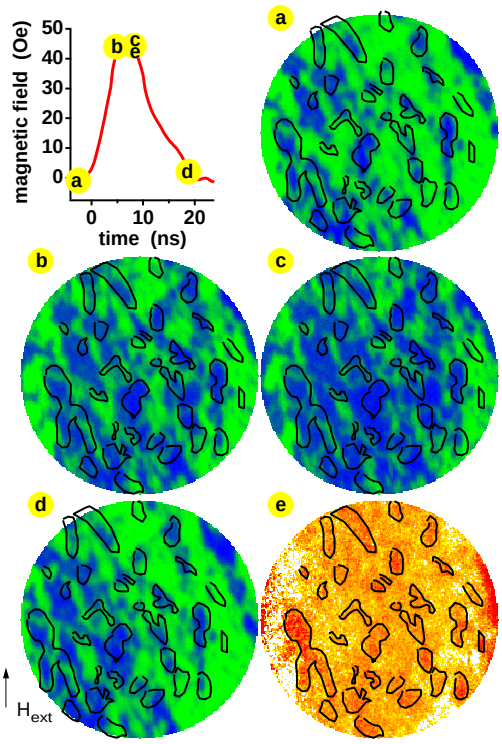

Fig. 2: Layer-resolved stroboscopic magnetic domain images of an $Fe_{20}Ni_{80}$/Cu/Co spin valve at different times during the periodic application of 18 ns-long magnetic field pulses (red curve). Panels a–d represent the magnetic domain pattern of the $Fe_{20}Ni_{80}$ layer at the time delays indicated in the graph, panel e represents the domain pattern of the magnetically harder Co layer. Black lines in the images indicate the position of domains in the Co layer.

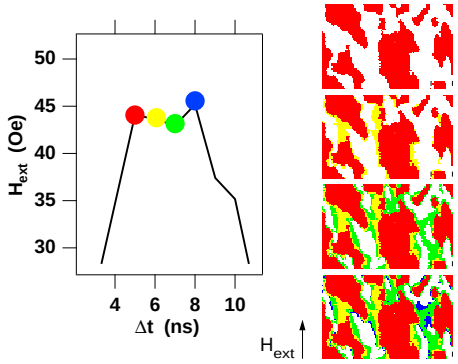

Fig. 3: Analysis of magnetic switching of the $Fe_{20}Ni_{80}$ layer by expansion of domains during the plateau of the field pulse. The images at the right represent the expansion of the switched areas in one nanosecond time steps. Differently colored areas indicate the increase of switched domains at the times indicated by symbols of the corresponding color in the graph on the left.

domains through propagation of domain walls (panels a–c). Upon reduction of the field, magnetic coupling to the Co layer leads to a shrinking of the blue domains (panel d), until after 1.6 μs eventually the starting configuration is restored (panel a), and the same cycle starts over again.

Closer analysis of the domain wall motion of the $Fe_{20}Ni_{80}$ layer during the plateau of the field pulse (between panels b and c of Fig. 2) is shown in Fig. 3. The right hand side shows the evolution of the $Fe_{20}Ni_{80}$ domain pattern in time steps of 1 ns. Different colors represent the incremental expansion of regions of switched magnetization at the times indicated in the graph at the left hand side. Due to the domain wall energy, which opposes the expansion of small domains, the domain wall velocity is greatest in areas where two existing domains merge together [3]. The magnetization reversal dynamics depends also on the competition between local magnetic interlayer coupling and intrinsic properties of the sample [4]. The combination of temporal, spatial and layer resolution makes this technique so extremely powerful for studying magnetization dynamics in layered magnetic systems.

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Monoatomic cobalt chains on Pd(110)

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In the last decade, nano-scale magnetic materials have been widely discussed due to the current development of electronic devices and data storage media. Thus it has been a particularly attractive topic to investigate the fabrication and magnetic properties of nano-scale, even atomic scale magnetic materials, such as nanowires or nanostripes, nano-particles or nano-dots, and nano-pillars. With the development of ultra-high vacuum and molecular beam epitaxy techniques, a rich variety of far-from-equilibrium nano-structures can be created depending on the atomic nature of the adlayer-substrate system, deposition temperature and -rates. Among them, the symmetry of the substrate is an important factor.

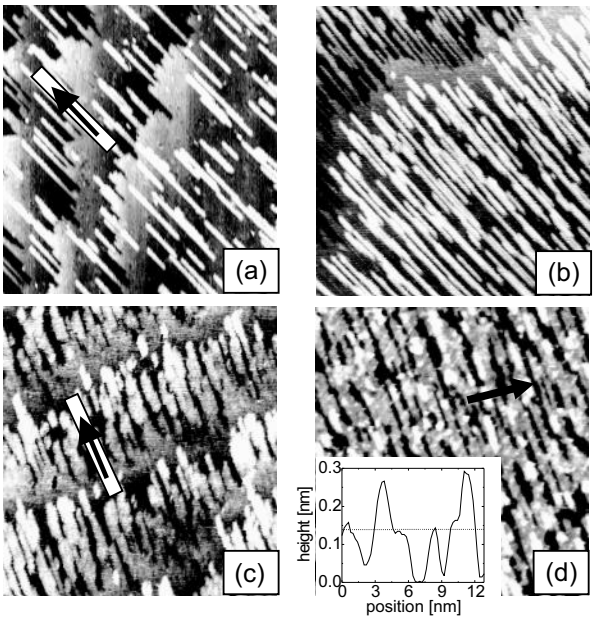

Fig. 1: STM images (50nm x 50nm) at different Co coverage on Pd(110) surface. (a) 0.15ML, (b) 0.3ML, (c) 0.5ML, (d) 0.65ML. STM line profile taken along the black arrow in (d) is shown in the inset, in which the horizontal line indicates the height level of the first Co atomic layer.

In this respect, the Pd(110) surface is an attractive example. The Pd(110) surface is one of the few unreconstructed fcc(110) surfaces which consist of close-packed atomic rows along the [1-10] direction separated by deep channels [1]. At not too high temperatures, preferential diffusion of incoming adatoms takes place along these channels, i.e., the [1-10] direction, resulting in island elongation, even formation of linear 1D chains along the (1-10) channels [2].

Co films were prepared by thermal deposition from a Knudsen cell. In order to suppress gas adsorption and interdiffusion of Co and Pd(110) substrate, the films were prepared at room temperature (RT), and then immediately transferred to the cooled sample stage for the magneto-optic Kerr effect (MOKE) measurements.

Fig.1 displays a series of scanning tunneling microscopy (STM) images characterizing the growth of submonolayer Co on Pd(110) at RT [3]. At the coverage of 0.15ML, Co monoatomic chains are formed and oriented along the [1-10] direction (Fig.1a). The monoatomic width of Co chains has been inferred from the topographic lines profile perpendicular to the chains which show discrete maxima at multiples of an elementary distance a=(0.4 ± 0.05)nm (corresponding to the spacing between two neighboring Pd atomic rows on the (110) surface). The fact that we were able to measure the lateral island separation down to two elementary distances indicates that the chains do not consist of double rows. With increasing coverage to 0.3ML, the average distance between Co atomic chains decreases, and some individual Co chains can span a whole terrace length (Fig.1b). At the coverage of 0.5ML, as shown in Fig.1c, Co atomic chains grow up to Co two-dimensional nano-islands which are preferentially elongated along the [1-10] direction with a width of 1.2nm-2.0nm. Moreover, the STM images clearly show the role of substrate step edges as heterogeneous nucleation centers. Both Co atomic chains and nanowires easily grow outward from substrate steps though Co nanowires are not aligned along the steps. When the coverage exceeds

0.6ML (see Fig.1d), the gaps between some Co islands are filled up, and the second layer nucleation is observed on top of the first Co layer. As demonstrated by the line profile of the inset, the height of the second partially completed layer of the Co islands has almost the same height as the first layer, i.e., one Co atomic layer. Moreover, the two layers thick Co islands are also exclusively oriented along the [1-10] direction.

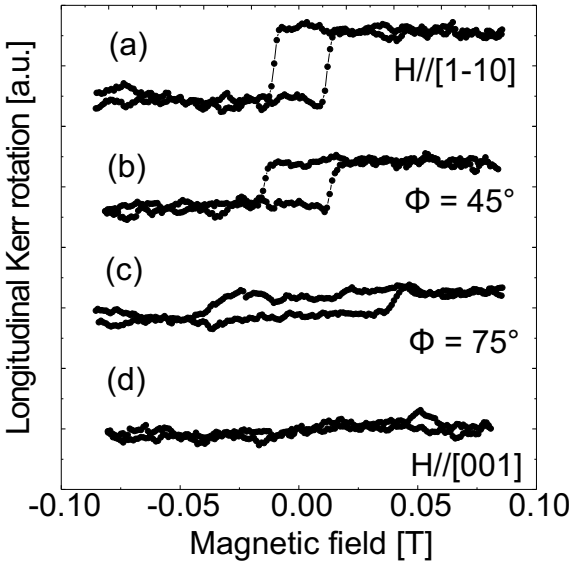

Fig. 2: Hysteresis loops of 0.5ML Co on Pd(110) surface measured along the [1-10] direction (a), under the angle $\phi = 45°$ (b), $\phi = 75°$ (c), and along the [001] direction (d).

An important issue for the investigation of Co nanostructures on Pd(110) is their uniaxial magnetic anisotropy. For less than 0.5ML Co coverage, it is difficult for our MOKE setup to detect Kerr signals. This is due to the lowest temperature of 50K that can be achieved, which is not enough for the ultrathin films of low Curie temperature (which is expected at the coverage below 0.5ML). Thus, Fig.2a-d show a series of representative MOKE hysteresis loops of 0.5ML Co on Pd(110) with the magnetic field applied along the different in-plane directions of Pd(110) surface. The magnetization is always probed along the field direction. The measurement temperature is 55K. The square hysteresis loop with the magnetic field applied along [1-10] direction displays minimum coercivity and maximum Kerr signal (see Fig.2a). When the magnetic field is applied in the sample plane under a variable angle from [1-10] direction (see Figures 2b and c), we found the intensities of the measured Kerr signals decreasing with increasing deviation angle of the magnetic field from the [1-10] direction. Simultaneously, we found the coercivity to increase with increasing deviation angle showing that a higher field has to be applied to reverse the magnetization. This fact indicates that the magnetization is not saturated along the direction of the applied magnetic field, and the measured Kerr signal is due to a projection of the magnetization oriented along the easy axis (i.e. [1-10]) to the magnetic field direction. With increasing deviation angle higher fields have to be applied to result in the same field component along the [1-10] easy axis. Moreover, the shapes of the Kerr loops are all almost square. This indicates that the magnetization direction is almost fixed to the [1-10] direction, i.e. the direction the cobalt nanowires are oriented along. Only at higher fields the magnetization rotates towards the field direction and finally saturates.

When the thickness of the Co film exceeds 1ML, the easy magnetization direction switches in plane to [001], i.e., is perpendicular to the nanowire direction. No polar Kerr signal corresponding to out-of-plane magnetization is observed. The polar signal appears only after the sample is: (1) exposed to the residual gas atmosphere for a couple of hours, (2) covered with protective layer e.g. of Au or (3) annealed above 400K. In particular, covering with Au results in the perpendicular anisotropy up to a thickness of about 6ML.

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Topologically induced magnetic frustrations in thin Mn films on Fe(001)

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In many magnetic devices, antiferromagnets in direct contact to ferromagnets play an essential role. Such combined systems allow the creation of artificial structures in which the magnetic properties can be tuned to technological needs. Fundamental properties concerning the interplay between ferromagnets and antiferromagnets are, however, not fully understood. In the last years, spin-polarized scanning tunneling spectroscopy [1] and spin-polarized scanning tunneling microscopy [2] became powerful tools for investigating magnetic structures on the nanometer scale. Even surfaces of antiferromagnets can be probed with these methods. Here, spin-polarized scanning tunneling microscopy was applied to study the behavior of magnetic frustrations at the surface of thin antiferromagnetic Mn films which are in direct contact to a ferromagnetic Fe(001) substrate [3].

To investigate the in-plane component of the spin polarization at the sample surface, we use ferromagnetic rings as electrodes [4]. A photo of a typical ring is shown in Fig. 1. The arrow indicates the approximate position at which tunneling occurs. A Ta wire is used to fix the ring to the scanner. A small alternating current is applied to the coil wound around the ring to switch the magnetization of the ring. A fast switching of the ring magnetization (switching frequency is much higher than the cut off frequency of the feedback loop) allows the separation of the topographic and magnetic information and the simultaneous imaging of both.

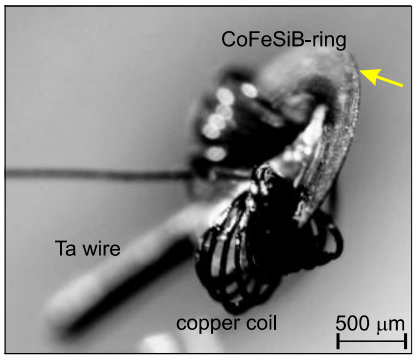

Fig. 1: Ring electrode as used for in-plane spin-polarized scanning tunneling microscopy measurements (outer diameter: 2 mm, inner diameter: 0.7 mm). Beside the ring, a small coil around the ring and a Ta wire are visible.

Mn on Fe(001) is a topological antiferromagnet as schematically shown in Fig. 2. Thus, adjacent Mn layers exposing at the sample surface are oppositely magnetized. Where Mn overgrows a monatomic Fe substrate step edge, the thickness of the Mn film on both sides of the step edge differs by one atomic layer. Buried Fe step edges appear as steps of sub atomic height at the surface of the Mn film because the out-of-plane lattice constant of Mn is slightly higher than the one of Fe(001), as indicated in Fig. 2.

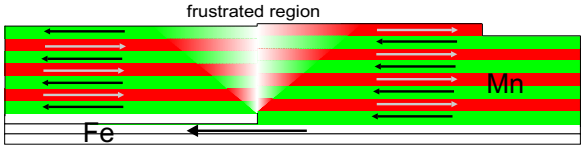

Fig. 2: Schema of the topological and magnetic behavior of Mn on Fe(001). The direction of magnetization of the ferromagnetic Fe substrate and the antiferromagnetic Mn film is indicated by the arrows and the different colors. Above a buried Fe step edge, a magnetically frustrated region is formed in the Mn film.

The magnetic situation above a buried Fe step edge is complicated. It is not possible to obtain simultaneously the same magnetic order at the interface between Fe and Mn on both sides of the Fe step edge and the antiferromagnetic order in the Mn film and the ferromagnetic order in the Fe substrate. Therefore, a magnetic frustration has to occur in the system. For thin films, it is likely that the frustration is present within the film, as schematically presented in Fig. 2. Mn layers with opposite magnetization directions are in contact above the Fe substrate step edge which results in a magnetically frustrated region in the Mn film.

Fig. 3a) shows the topography of 11.9 ML Mn on Fe(001). Islands on a closed Mn layer which are separated by monatomic Mn steps are visible. A step of subatomic height is running almost vertically through the image (Fig. 3a)), indicated by the black arrows. This step is created by a buried Fe substrate step edge (Fig. 3c)). The alternating contrast in the spin signal shows the layer-wise antiferromagnetic order between the three Mn layers (Fig. 3b)). The islands are coupled antiferromagnetically to the layer below them. Beside the antiferromagnetic order between the islands and the closed Mn layer, a reversal of the contrast appears above the buried Fe step edge. This means, the magnetization above the Fe step edge rotates by $180°$. Therefore, indeed a magnetically frustrated region is created in the antiferromagnetic Mn film caused by a topological defect, a monatomic step at the interface. A line profile across the magnetic frustration shows that its width extends over several nm (Fig. 3d)). Spin-polarized scanning tunneling microscopy is surface sensitive and thus, does not allow a direct investigation of the magnetic order within the Mn film. It is likely, however, that the magnetic frustration starts at the interface and extends throughout the entire film as schematically shown in Fig. 2.

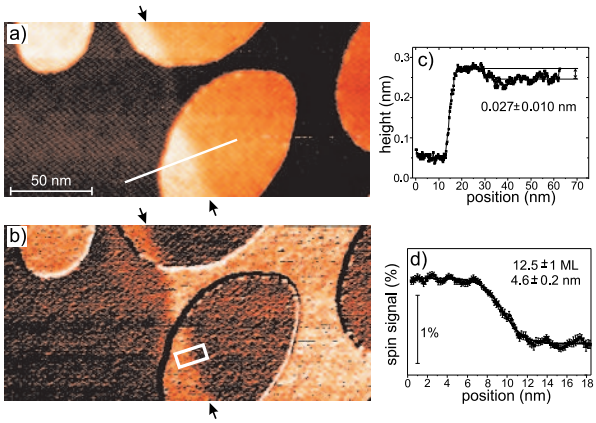

Fig. 3: Spin-polarized scanning tunneling microscopy image of a) the topography and b) the corresponding spin signal of 11.9 ML Mn on Fe(001). One buried Fe step edge is visible, indicated by the arrows. The line profile in c) shows a monatomic Mn step and a step of subatomic height caused by a buried Fe step edge. The line profile is taken along the line in a). d) Line profile across the magnetic frustration in the Mn over-layer at the position of the box in b). The solid line represents a fit to the measured data.

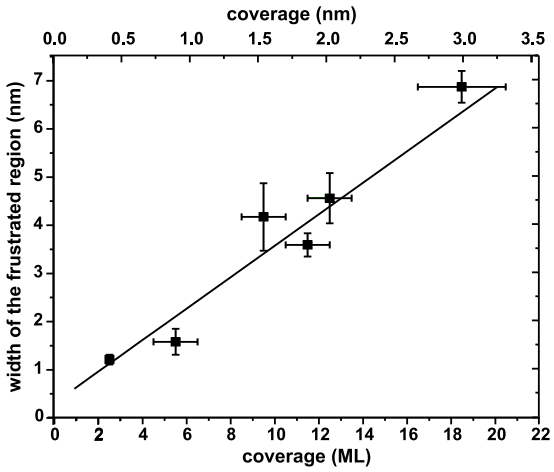

Fig. 4: Width of magnetically frustrated regions, measured at the Mn surface layer as a function of the Mn film thickness in ML (bottom scale) and equivalent in nm (top scale). The solid line is a linear fit to the experimental data points.

We investigated the width of the magnetically frustrated region as a function of the Mn film thickness. The spin-polarized scanning tunneling microscopy measurements showed that the magnetic frustration widens linearly with increasing Mn film thickness up to about 20 ML (Fig. 4). Because of a phase transition of Mn at about this film thickness, thicker films could not be investigated. The observed linear widening indicates that the width of the magnetically frustrated region is still much smaller than the equilibrium width in bulk Mn.

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Lattice Relaxation and Spin-Reorientation in Fe/Ni/W(110) Trilayers

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The preferred direction of the magnetization (”easy axis”) in a magnetic material is a fundamental property thoroughly studied in basic research. It is of primary importance for applications also. Much interest is devoted to manipulate the easy axis to point in a certain direction. A prominent example is the development of high density magnetic storage devices based on alloys characterized by the easy axis magnetization perpendicular the film plane.

The easy magnetzation direction is governed by the magnetic anisotropy energy, which is commonly discussed in terms of bulk (”magnetocystalline anisotropy”) and the surface (”surface anisotropy”) contributions.

A stunning example for the subtle effects which govern the magnetic anisotropy energy is the adsorbate-induced switching of the the easy magnetization direction. Due to the lack of precise structural data, the reorientation transition of the magnetization direction is tentatively attributed to changes of the surface contribution to the magnetic anisotropy energy, while the influence of adsorbate-induced structural changes is neglected.

This approach is a priori not well justified. The reason is that lattice strain and magnetic anisotropy are intimately connected by the the so-called magnetoelastic coupling. The strength of this coupling is orders of magnitude larger as compared to the magnetocrystalline anisotropy, and even subtle structural changes may result in a spin reorientation transition [1].

We performed combined magneto-optical Kerr effect and surface x-ray diffraction experiments at the European Synchrotron Radiation Facility (ESRF, Grenoble) to study the impact of adsorbate-induced structural changes on the magnetic anisotropy. This unique combination of techniques is a necessary prerequisite to study magnetic and structural properties and their change upon adsorption in situ.

It was the goal of our work to investigate, whether the spin reorientation transition of Ni(111) layers from in-plane to out-of-plane magnetization upon Fe-adlayer coverage is accompanied by a respective structural change. Calculations show that a tiny change of the average layer spacing of the Ni stack of 0.015 Å would already be sufficient to drive the spin reorientation stransition [1]. Our combined in-situ magnetization and surface x-ray diffraction experiments indicate, that the change of the average film layer spacing $\Delta d_{(111)}$ is below 0.002 Å when the adlayer-induced change of the easy magnetization axis from in-plane to out-of-plane occurs. Thus, our results provide direct experimental evidence that the reorientations of the easy magnetization axis cannot be attributed to the effect of lattice strain changes, but are induced by adlayer-induced changes of the surface anisotropy energy.

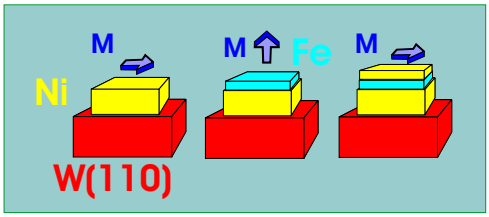

Fig. 1: Schematic view of the Fe/Ni/W(110) system with the easy magnetizaton directions indicated by the arrow. Adsorption of one atomic layer of Fe leads to out of plane magnetization (center), while another layer of Ni reverts the magnetization back to the plane (right)

The changes of the preferred magnetization axis in the Fe/Ni/W(110) adsorption system are schematically displayed in figure 1. For deposition of more than 3 layers of Ni on

W(110) the magnetization is in the plane (see fig. 1, left).

The preferred magnetization direction changes to the surface normal, when one atomic layer of Fe is adsorbed (center). It is reverted back to in-plane by deposition of another layer of Ni-atoms (right).

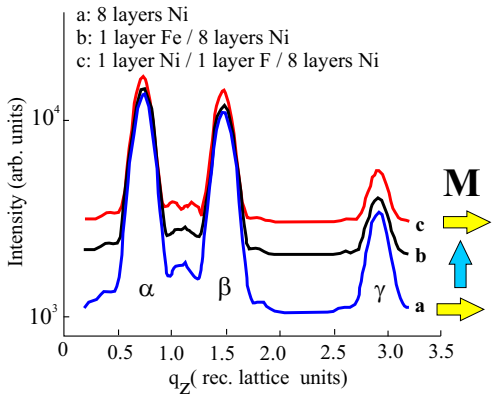

Fig. 2: Scans along q_z shown on a logarithmic scale after deposition of 8 atomic layers Ni on W(110)(a) and after subsequent deposition of 1 layer Fe (b) and 1 layer Ni (c). Peaks (α) and (γ) are related to the (101) and (104) reflection of the fcc-like structure, peak (β) is related to the crystallographic twin. Arrows schematically indicate the magnetization direction. Curves are shifted vertically for clarity.

While in situ magneto-optic Kerr effect experiments confirmed the changes of the preferred magnetization direction from in-plane to out of plane and back, surface x-ray diffraction data did not indicate changes of the vertical lattice spacing larger than 0.002 Å. This accuracy is about one order of magnitude better than achievable by structure analysis techniques on an absolute basis.

Figure 2 shows surface x-ray diffraction measurements along the sample normal probing the vertical lattice spacing of the film. Scans were measured after deposition of 8 layers Ni (a), and subsequent deposition of one layer Fe (b), and another layer Ni (c). In total, three Bragg-peaks are observed, where the first (α) and third (γ) can be indexed as the (101) and the (104) reflections of the fcc-like structure satisfying the condition $-h + k + l = 3n$ (n=integer), while the second (β) represents the reflection from the crystallographic twin related to the reversed layer stacking (ACB versus ABC).

Both, direct qualitative inspection of the peak positions and profiles as well as their quantitative analysis indicate that there are no differences between the experimental curves within experimental resolution. Adsorption of one adlayer does not lead to a shift of the peak positions larger than 5×10^{-4} to 1×10^{-3} reciprocal lattice units [1]. This corresponds to an upper limit for the lattice spacing change of 0.002 Å.

The calculated lattice relaxation required to induce the switching of the easy axis from in-plane to out-of plane (Δd=0.015 Å) is roughly one order of magnitude larger. Therefore, we have direct experimental evidence that magnetoelastic contributions due to changes of the lattice geometry cannot be accounted for the adsorption induced switching of the easy axis but must be attributed to a change in the surface (interface) anisotropy energy.

While this study is focused on Fe/Ni/W(110), we expect that our novel experimental approach to combine in-situ magnetic measurements with relative measurements of the inter-layer relaxation is relevant for thin film magnetism in general. Since (adsorbate-induced) spin reorientation transitions are a common phenomenon, but precise structural data are generally lacking, accurate structure determination are clearly called for. The highly accurate structure data offer a reliable reference for state of the art calculations of magnetic anisotropy energy [2]. Our approach points to a solution of the long standing problem how to unravel the different anisotropy contributions which drive a spin reorientation transition.

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Strain-Induced Adatom Motion on Metal Surfaces

V.S. Stepanyuk and J. Kirschner

Progress in atomic engineering makes it now possible to produce the two-dimensional magnetic nanostructures tailor-made on the atomic scale[1]. Advanced fundamental studies on magnetic nanostructures on metal surfaces indicate that not more than a few hundred atoms might be needed to achieve magnetic terabit memories[2]. As device miniaturization shrinks into the mesoscale, the control of the individual atomic events involved in the formation of magnetic nanostructures becomes crucial. Here we demonstrate new unexpected effects in atomic motion on metal surfaces.

In the early stages of growth atomic diffusion and interactions between adatoms lead to the formation of stable and metastable small islands. Atomic diffusion in the vicinity of the islands governs their size, shape, density and finally the growth mode. It is usually assumed that at low island densities a two-dimensional random motion of isolated adatoms or tracer diffusion takes place. The corresponding diffusion coefficient is given by $D = \frac{1}{2d} n \langle l^2 \rangle \nu_0 exp\{-\frac{E_m}{k_B T}\}$, where $\langle l^2 \rangle$ is the mean square hop length, d is the dimensionality of the motion ($d = 2$ at the surface), n is the number of equivalent hop directions (for fcc(111) $n = 3$) and ν_0 is the attempt frequency or prefactor. Experimentally parameters E_m and ν_0 are determined by analysing the dependence of island density on temperature and the deposition flux in the framework of the classical nucleation theory. However, we have revealed that the atomic motion near small islands and the migration barrier on metal surfaces with a weak corrugation are drastically influenced by the mesoscopic relaxation of islands and substrate[3].

Performing atomic scale calculations we have found a strong oscillation of adatom migration barrier near the Co islands on Cu(111) for different island sizes [3]. The results shown in Fig.1 demonstrate that the island and the surface layers are not flat due to the mesoscopic relaxations at the interface. The island induces anisotropic strain in the substrate which makes adatom motion parallel to the island side more preferable than in the direction to the island. The strain originates from existence of the size-dependent mesoscopic mismatch between the island and the substrate [4] and makes the migration barriers also dependent on the size of the islands.

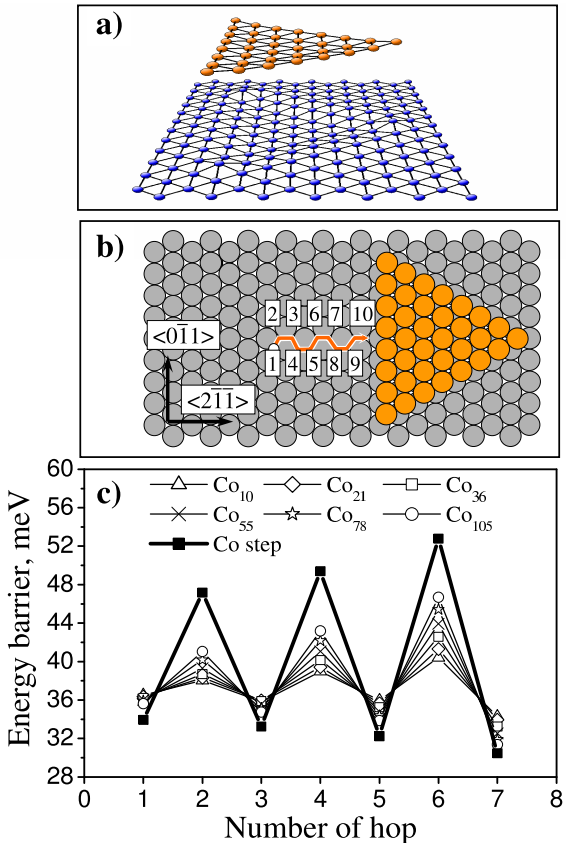

Fig. 1: (a) The shape of Co_{36} island and the Cu substrate.(b) Schematic view of Co_{36} island and the diffusion path of Co adatom. Numbers denote the hops of the adatom. (c) Comparison of migration barriers for Co adatom near triangular Co islands of different sizes and near Co step.

A sample diffusion path of Co adatom towards Co island is schematically shown in Fig.1b. Numbers denote the elementary hops

of the adatom along the $\langle 2\bar{1}\bar{1} \rangle$ direction. Energy barriers corresponding to the hops are plotted as a function of the adatom - island separation. As the adatom approaches the island side, the migration barrier exhibits strong oscillatory changes. The barrier oscillations become stronger for larger islands (Fig.1c). In the limit of step the amplitude of the oscillations is more than two times larger than for the Co_{36} island.

Another example which demonstrates the impact of mesoscopic relaxations on atomic motion is interface intermixing. It has been believed that intermixing at metal interfaces occurs mainly due to an atomic exchange of single adatoms with the substrate atoms. However, our studies have shown a complete failure of this assumption. As an example, we discuss results for the Fe/Cu(001) interface. For the simple atomic exchange between Fe and Cu atoms we obtain enegy barrier of about 0.78 eV. However, the barrier for the exchange is reduced to 0.6 eV near the Fe dimer embedded in the topmost layer. In Fig.2a we show the atomic displacements of the Cu substrate near the embedded Fe square island (color thermometer). These results indicate that strain relaxations in the embedded islands can lead to pronounced structural changes in the substrate. Our results have revealed strong displacements of the substrate atoms in the direction towards the island, which have been found to strongly depend on the size of islands. The exchange barriers for the Fe adatom near small embedded islands of different sizes are shown in Fig.2b. One can see that the exchange barrier strongly depends on the size of islands and the distance between the adatom and the island. We have revealed that the size-dependent relaxations in islands and the substrate cause these unexpected effects. Our studies offer a consistent explanation of the mechanism of intermixing in metal heteroepitaxy. This a collective process where adatoms tend to exchange their positions with the substrate atoms near small embedded clusters.

None of the above effects could be predicted by the classical concept of monolayer growth. The driving force for the strain-mediated adatom motion on metal surfaces is associated with the size-dependent mesoscopic relaxations in small islands.

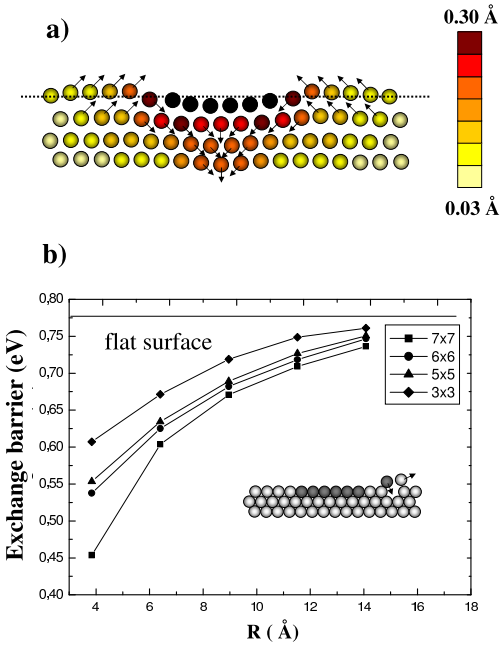

Fig. 2: (a) The shape of the embedded Fe island of 36 atoms and the Cu substrate; the color marks the magnitude of the total displacement. (b) Exchange barrier for the Fe adatom on the Cu(001) surface near embedded square Fe islands. The horizontal line represents the exchange barrier for a flat surface.

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Metal nanotube membranes and their applications

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Two-dimensional (2-D) periodic arrays of metal or semiconductor nanodots have attracted a lot of research attention recently, due to their potential applications, e.g., in high density data storage media, electronic/optoelectronic devices, miniaturized biochips or sensors. Various approaches have been employed to realize well-defined 2-D arrays of nanodots on a large scale [1]. State-of-the-art lithographic techniques could provide a good opportunity for controlling feature size, shape, and spacing. However, the available conventional lithographic techniques have a fundamental drawback in terms of cost, scalability, and process time.

We report a simple and general method for fabricating hierarchically organized arrays of Au nanotubes in the form of membranes, which can be used as a shadow mask in sputter deposition and focused ion-beam (FIB) patterning for generating spatially well-resolved 2-D periodic arrays of well-defined nanostructures on a cm^2 scale. The present nanofabrication method uses an electro-chemical deposition technique to replicate a master pattern structure of a nanoporous anodic aluminum oxide (AAO). By taking advantage of the easy adjustment of the pore diameter, we could fabricate structurally well-defined shadow masks having different aperture sizes, which provide a good opportunity for controlling the feature sizes of nanostructures in sputter deposition and patterning by reactive ion etching (RIE).

Free-standing Au nanotube membranes were fabricated as follows. First, a thin metallic gold film was sputtered onto the surface of the AAO template. This process resulted in a thin conducting metal layer on the top part of the inner nanochannel surface as well as on the top surface of the AAO template. Subsequent electrochemical deposition of Au homogeneously thickens the conducting metal layer, resulting in a tubular metallic nanostructure inside the alumina nanochannels. After removal of the alumina template in phosphoric acid, the resulting gold membrane was floating on the surface of an etching solution. After a cleaning procedure the Au nanotube membrane was transferred onto any kind of substrates from the solution [2].

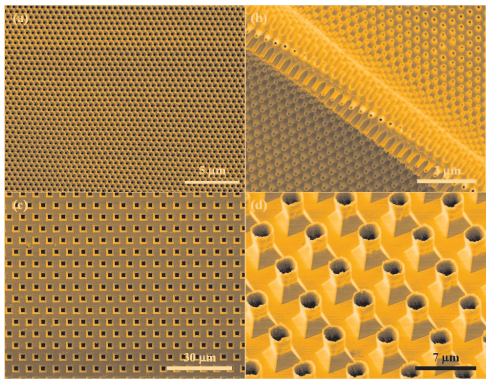

Fig. 1: SEM images of Au nanotube membranes replicated from (a, b) nanoporous AAO and (c, d) macroporous silicon.

Fig. 1 shows some representative SEM images of the Au nanotube membranes that have been fabricated from an as-prepared AAO template (a, b) and macroporous silicon (c, d). The two sides of the Au nanotube membranes are not equivalent. The structure of the front side (Fig. 1a) is characterized by a close-packed hexagonal arrangement of nanopores with a center-to-center distance of 500 nm. The other side of the Au nanotube membrane has a different surface morphology with a uniform array of Au nanotubes, as shown in Fig. 1b. Application of the present mask fabrication technique can also be extended to other porous templates such as macroporous silicon or patterned silicon substrates with lithographically generated etch pits (see Fig. 1c-d), demonstrating the general applicability of the process for fabricating various metallic masks.

The fabricated Au nanotube membranes were used as shadow masks for generating extended 2-D arrays of metal nanodots *via* sputter deposition. Figure 2b shows typical SEM images of Pt nanodot arrays sputtered on silicon substrate using the Au nanotube membranes shown in Figure 2a as a shadow mask. As evident from the presented SEM images, each nanodot is spatially well-resolved with perfect hexagonal arrangement. The demonstrated process provides a good control over the size of nanodots. The size of the deposited nanodots can easily be tuned, independent of the interdot distances.

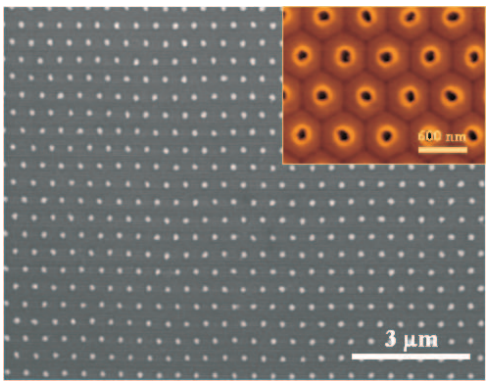

Fig. 2: SEM image of Pt nanodot arrays sputter deposited on Si substrate using Au nanotube membranes as a shadow mask (inset).

In our experiments, 2-D arrays of metal nanodots can be routinely generated with a dot diameter from 85 nm to 250 nm. It is expected that such a unique capability of tailoring the dot sizes independent of the interdot distance will enable us to systematically investigate the magnetic coupling between neighboring magnetic nanodots, after deposition of magnetic materials.

Alternatively, the Au membranes can be used as etch masks to transfer a regular pattern of holes to a substrate. This is demonstrated in Fig. 3a, where a hexagonal lattice of holes of 100 nm diameter was created in 6H-SiC(0001) by reactive ion etching (RIE). While the position and the average diameter of the holes is well transferred, the individual shape of the holes slightly varies. In a post-processing step of etching with hydrogen at elevated temperatures [3], the slightly irregular shape of the holes may be improved to almost perfect hexagons (see Fig. 3b) reflecting the anisotropy of the surface free energy of the SiC crystal. Each individual hole consists of an outer set of six facets tiled by 24 ° from the surface plane surrounding the inner hexagonal hole with perpendicular walls [3]. Regular hole arrays of hole aspect ratios between 2 and 3 could be created this way.

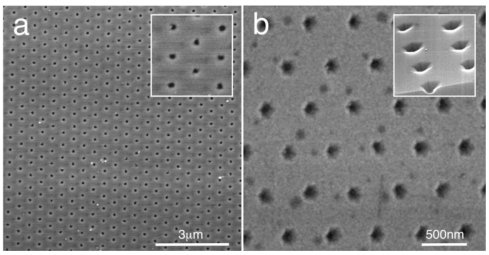

Fig. 3: SEM images of the nanostructured 6H-SiC(0001); (a) before and (b) after hydrogen etching at 1500 ^{o}C.

In summary, this report describes a novel method for fabricating well-defined metallic membranes. The nanotube membranes have been successfully used as a shadow mask not only in sputter deposition for the generation of two-dimensional nanodot arrays of various material but also in reactive ion etching (RIE) and for the pattern transfer to semiconductor substrates (e.g. SiC). Apart from the applications described the growth of 2-D perfect hexagonal arrays of ZnO nanorods on large scale has also been demonstrated, by combining the well-established vapor-liquid-solid (VLS) growth process for semiconductor nanowire and this mask technique. Furthermore, the metal nanotube membranes with a sub-wavelength scale aperture exhibit superior optical properties, which are under current investigations.

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Three-Dimensional Silicon-Based Photonic Crystals Fabricated by Electrochemical Etching

S. Matthias and F. Müller

Large scale, highly periodic structures have gained considerable interest in a number of areas in modern physics serving as photonic crystals [1, 2], sensors or as massively parallel Brownian ratchets [3]. Recently photonic bandgap materials became increasingly important in which three-dimensional periodic variations of the dielectric constant controllably prohibit electromagnetic propagation throughout a specific frequency range and inhibiting spontaneous emission [2]. Realizing the theoretically proposed unusual electromagnetic properties, perfectly structured dielectric materials on the length-scale of the wavelength of light are required.

This report focusses on the accurate and fast electrochemical fabrication of three-dimensional microstructures with nanometer precision for photonic crystal applications, in particular a simple cubic arrangement of overlapping air-spheres possessing a complete 3-dimensional bandgap [4].

Macroporous silicon is grown in a photo-electrochemical etching process [5]. An n-type (100) oriented silicon wafer is prepatterned by standard photolithography. Subsequent alkaline etching forms inverted pyramids acting as initial pores. Under anodic bias and backside illumination the wafer is then etched in hydrofluoric acid. The photo-generated minority carriers diffuse through the whole wafer. The silicon-electrolyte contact reminds of a Schottky-contact, where a space charge region within the silicon evolves. The applied potential enlarges the width of the space-charge region (SCR). The SCR acts on the electronic holes and concentrates them mainly on the pore tips, promotes the dissolution of silicon and results in a pore growth straight along the $\langle 100 \rangle$ direction. Because a n-type silicon wafer is used, the applied voltage and the etch current can be independently chosen. The area of the resulting air-pores is approximately proportional to the etch-current and can be varied during the growth by changing the intensity of the applied backside illumination [6]. However, this process has restricted the resulting pore shapes to smooth sinusoidal or ratchet-type ones with small variations in diameter so far. As a rule of thumb, the shorter the length of the modulation the lower are the variations of the diameter, which can be realized.

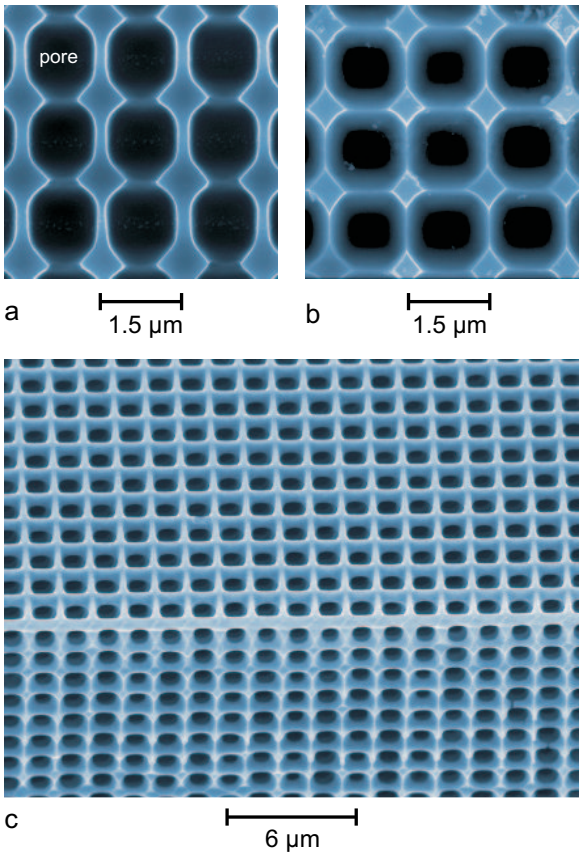

Fig. 1: Scanning electron micrographs of cleaved modulated macroporous silicon wafers **a**: 3D macropores are arranged in a square lattice with a pitch of 1.5 µm. **b**: After the isotropic widening procedure an almost simple cubic arrangement of overlapping air-spheres in silicon is obtained **c**: Widened structure form bird's eye view. On top the lithographically defined square lattice and on bottom the etched one.

Strong variations in diameter on a distance less than the lateral lattice constant as well as pore shapes with sharp edges are required

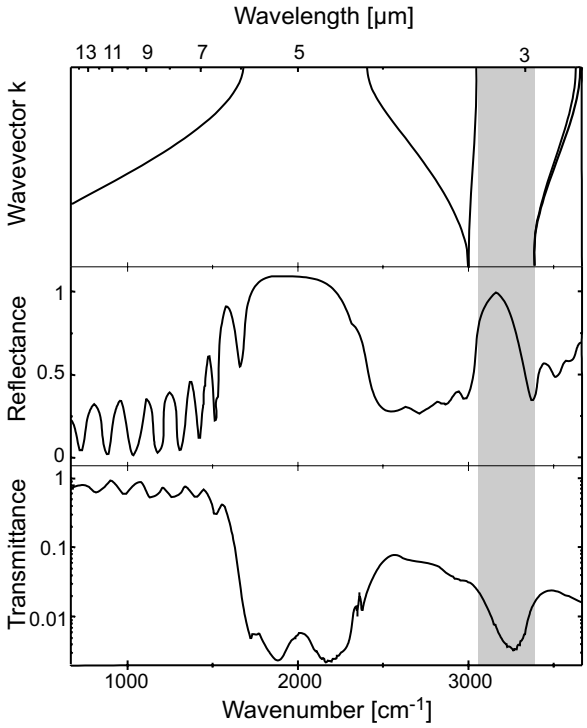

Fig. 2: Reflectance and transmittance measurements of the sample of fig. 1c along the growth direction. The obtaind spectra agree well with a bandstructure calculation. The grey highlighted region indicates the complete photonic bandgap.

for the fabrication of a simple-cubic photonic crystal of overlapping air spheres in silicon. The observed smoothing of the pore shape can be significantly reduced by increasing the applied voltage although this leeds to some instability in pore growth. Both, etch current and the applied voltage have to be modulated allowing to adapt the SCR continuously and to optimize the focussing effect for the holes during the etching process. With a carefully adjusted current and voltage profile the required sharp edges and large diameter variations can be grown (fig. 1a).

To obtain the cubic symmetry and the optimum porosity further post-processing of the porous sample is required. In a homogeneous and isotropic widening step the pore diameter is increased. First, the samples are annealed at 900°C for 150 minutes to grow a silicon oxide on the silicon surface. In a second step the oxide is removed by a hydrofluoric etching step. The widening may be repeated several times to precisely adjust the filling factor. Finally, this uniform erosion leads to interconnected pores also in the plane perpendicular to the pore axis (fig. 1b, c). Although starting from a columnar structure, the geometry obtained is very close to cubic overlapping air spheres in silicon.

The obtained samples were characterized with a Microscope Fourier Transform Infrared Spectrometer (FTIR). Figure 2 shows the transmittance and reflectance along the growth direction. Air and a silver mirror served as background respectively. The transmission through the photonic crystal is reduced by more than 2 orders of magnitude within the fundamental stop gap. In addition Fabry-Perot-Resonances resulting from the multiple reflection at the boundaries indicate a good quality crystal. The grey highlighted region indicates the complete 3-dimensional photonic bandgap.

The well-established photo-electrochemical etching process by Lehmann [5, 6], was significantly improved to allow the manufacturing of perfect three-dimensional microstructures. A simple cubic lattice of overlapping air-spheres results by a homogeneous widening of strongly modulated samples with sharp edges. Moreover, this process allows a precise control of the resulting shape with a resolution in the nanometer range.

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Silicon and SiGe nanowhiskers grown by molecular beam epitaxy

L. Schubert, N. D. Zakharov, G. Gerth, and P. Werner

Nanowhiskers of silicon and silicon/germanium are of increasing interest due to their specific physical properties as well as their potential for new nanodevices (1). They are normally grown by chemical vapor deposition (CVD) or by gas-source molecular beam epitaxy (GS-MBE). To initiate their formation, small droplets of metals, such as gold, are applied, which form low-temperature eutectic liquids with silicon acting as a seed for the whisker growth. The silicon is preferentially incorporated via the liquid silicon-metal droplet. This technique is referred to as the vapor-liquid-solid mechanism (VLS-mechanism) (2,3). It is characterized by a growth process at the whisker/droplet interface by the incorporation of Si atoms coming from the liquid droplet. However, this growth process is still under controversial discussion and might strongly depend on the specific techniques applied.

We applied molecular-beam epitaxy (MBE) as a growth technique, since it allows a complete in-situ growth of Si and Si/Ge nanowhiskers under reproducible conditions (e.g. clean oxide-free surfaces). Concerning the specific growth conditions we refer to (4). Fig. 1 shows a typical SEM image of such Si nanowhiskers with Au droplets on their tips. From corresponding experiments we determined the ratio between the diameter d and the length l of the whiskers as a function of the growth time and the growth temperature T_s. Different growth processes should yield a different *l/d* ratio. As an example, Fig. 2 presents the result of two experiments, where Si whiskers were grown at 525°C for 60 min and 120 min, respectively. In the case of MBE growth, for a given growth time the length of the whiskers is systematically longer for whiskers with smaller diameters (see Fig. 3), which is opposite to what has been found for CVD grown whiskers (2,3). A simple fit based on a power law $l = Cd^m$ leads to $m \approx -1$. C denotes a constant. In the growth model presented elsewhere (4) we

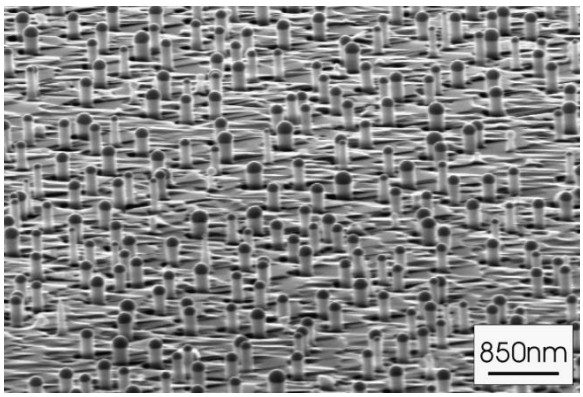

Fig. 1: SEM image of Si-whiskers grown on a (111) Si substrate at a Si growth rate of 0.5 Å/s for 120 minutes at a substrate temperature of 525°C.

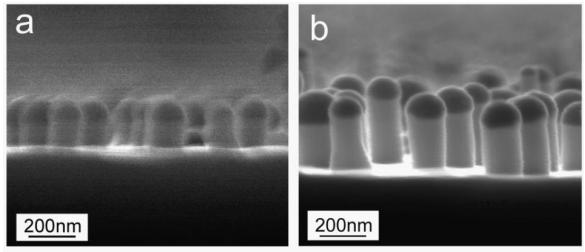

Fig. 2: SEM cross-section images of Si-whiskers grown on a (111) Si substrate at 0.5 Å/s for 60 minutes (a) and 120 minutes (b) growth time. In this specific experiments T_S amounted to 525°C.

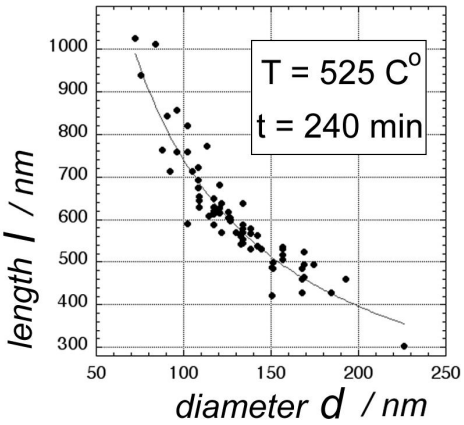

Fig. 3: Correlation between the length l and diameter d of Si whiskers. In this specific case the growth experiments were performed at 525°C for 240 min.

explain this behavior by a Si ad-atom diffusion on the surface of the whiskers and a reaction controlled incorporation of the Si atoms at the circumference of the droplet basis. This diffusion related component leads, for the case of MBE growth, to a larger growth rate for nanowhiskers with smaller radius in accordance to the observed radius dependence. For MBE-grown Si whiskers, the (111) growth interface between the whisker and the droplet is not plane, but shows side walls as seen in Fig. 4. This indicates that the incorporation

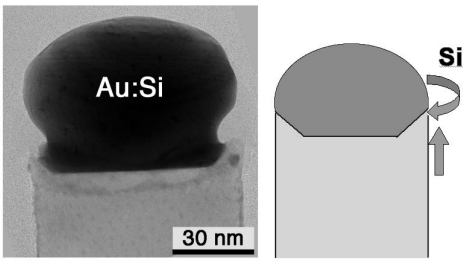

Fig. 4: Cross-section TEM image of the Si whisker tip (left) and the scheme of the interface structure. The whisker grows in $\langle 111 \rangle$ direction. The center is characterized by atomically flat $\{111\}$ planes.

of the Si ad-atoms occur at regions, where the three phases (V,L,S) are in common contact. From these walls $\{111\}$ planes can flow in horizontal directions.

Based on experiences we generated silicon/germanium heterostructures. The goal was to study the phenomena of the VLS growth behavior of Si whiskers including thin Ge layers in the nanometer-range (5). Fig. 5a shows a TEM cross-section image of such an experiment, where a whisker was grown at 545°C. Three Ge layers were incorporated with a nominal thickness of 0.5, 1 and 1.5 nm separated by a 25 nm thick Si layer. The Ge layers appear as dark lines. In the whisker the Ge concentration is slowly increasing within a layer of 10 to 15 nm (Fig. 5b). This might be caused by the slow increase of the Ge concentration in the Au/Si droplet after opening the shutter of the Ge source. Sharper Ge layers might be generated in thinner whiskers ($d < 70$ nm).

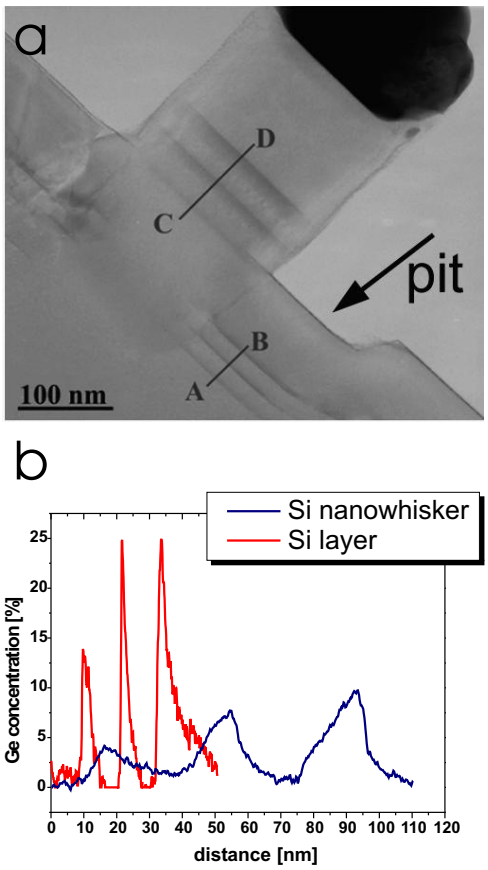

Fig. 5: TEM cross-section image (a) of a whisker where three Ge layers were incorporated. The TEM contrast of the Ge lines corresponds to the Ge concentration. b) shows the Ge concentration profile along the marked lines A-B and C-D, respectively.

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Diameter-dependent Growth Direction of CVD-grown Epitaxial Silicon Nanowires: Experiments and Modelling

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We report on experiments demonstrating that the growth direction of epitaxially grown silicon nanowires changes with decreasing nanowire diameter from <111> to <110> direction. A model is proposed to explain this phenomenon.

We investigated silicon nanowires that were grown epitaxially on a Si (100) substrate. The nanowires were produced by UHV chemical vapor deposition (CVD) using diluted silane as precursor gas and gold as catalyst. Growth takes place at a temperature of 400 °C and at a silane partial pressure of 10 Pa. Fig. 1 shows a typical top view SEM image of a sample. It exhibits three different kinds of nanowires. First, the ones pointing at ±18°

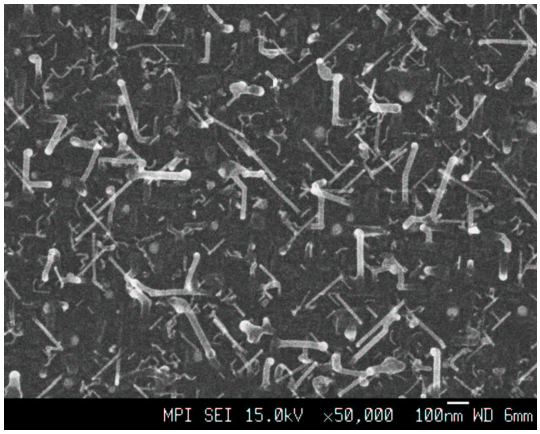

Fig. 1: Top view SEM image of nanowires grown on a Si (100) substrate.

with respect to the horizontal (x-axis) or vertical (y-axis) on Fig. 1. These nanowires are unambiguously <112> oriented. Second, thin nanowires growing at ±45° to the horizontal and vertical, respectively, which can be identified to be <110> oriented. Third, the ones observed horizontally or vertically in the projection plane. These could either be <211> or <111> oriented. The pure <111> distribution can be deduced from this mixed distribution by subtracting the <112> contribution. The resulting relative proportions of the different orientations are depicted in Fig. 2.

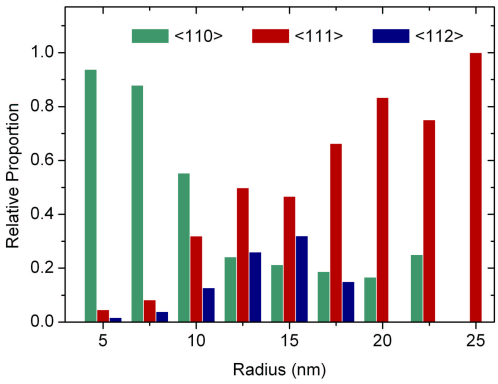

Fig. 2: Relative proportion of the different growth directions as a function of the radius.

Here it can be seen that for radii smaller than 10 nm the <110> direction is preferred, and that the <111> direction becomes dominant for radii greater than 15 nm. So the transition between the <111> and the <110> orientation takes place at a crossover radius r_c of 10-15 nm, which agrees with the results of Wu et al. [1] for non-epitaxially grown silicon nanowires. In the following, a model explaining this transition will be presented.

It is well known that CVD-growth of silicon nanowires can be described by the vapor-liquid-solid (VLS) mechanism [2], which means that silicon from the precursor gas is incorporated in the Au/Si-eutectic droplet and sweated out at the liquid-solid interface. Consequently, growth takes places only at the liquid-solid interface and it seems obvious to assume that the properties of this interface also determine the direction of growth. We assume that the free energy per circumference $f = F/L$ of a <111> oriented liquid-solid interface is dominated by two growth-direction dependent terms:

$$f = a\sigma_{ls}\, r + \Delta z \sigma_s \,. \qquad (1)$$

First, the contribution from the interface energy itself, which is proportional to the interface tension σ_{ls}, to the radius r of the nanowire, and to the dimensionless constant

a, defined as $a = A\,L^{-1}r^{-1}$, with A being the interface area. In the case of a <111> oriented nanowire having a regular hexagonal cross section, we define the radius r as the center-to-corner distance of the hexagon. As for the <110> oriented nanowires, the definition of r is more complicated, because their hexagonal cross section is not regular. So in this case we define the radius of the nanowire as the center-to-corner distance of a regular hexagon of equal circumference.

The second contribution to f comes from the edge of the liquid-solid interface. This term is proportional to the surface tension σ_s of the wire surface and to the interface thickness Δz, as indicated in Fig. 3. The free energy f, as given by (1), is schematically depicted in Fig. 4 as a function of r. The y-axis intercept is given by $\Delta z\sigma_s$ and the slope is proportional to $a\sigma_{ls}$. The interface energy

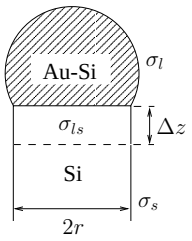

Fig. 3: Schematic of the nanowire tip.

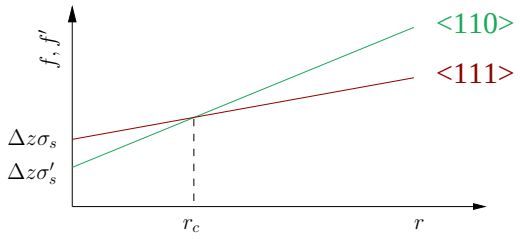

Fig. 4: Free energy per circumference of both <110> and <111> oriented nanowires.

f' of a <110> oriented nanowire can be expressed similarly, with the only difference that the slope and the y-axis intercept are now determined by the corresponding quantities a', σ'_{ls}, and σ'_s of a <110> oriented nanowire. If $\sigma_s > \sigma'_s$ and $a'\sigma'_{ls} > a\sigma_{ls}$, the two functions f and f', are intersecting at the crossover radius

$$r_c = \Delta z\,\frac{\sigma_s - \sigma'_s}{a'\sigma'_{ls} - a\,\sigma_{ls}}\,. \qquad (2)$$

In Fig. 4 one can see that minimizing the free energy with respect to the growth direction results in <110> oriented nanowires for radii smaller than r_c. For radii greater than r_c, the <111> direction is energetically more favorable.

An estimate for the crossover radius r_c can be obtained if the parameters a and a', the surface/interface tensions and the width Δz are known. The parameters a and a' can be found by considering the specific geometry of the interface area [3]. The surface/interface tension values can be deduced from references [4-6] if it is additionally assumed that the surface/interface tension is roughly proportional to the density of dangling bonds and that the surface of a <111> oriented nanowire consists of six $\{110\}$ planes. Furthermore, the interface thickness Δz is assumed to be $\Delta z \approx 1$ nm. Plugging all things into equation (2) then leads to an estimated crossover radius $r_c \approx 10$ nm. This value fits very well to the experimentally observed crossover radius as shown in Fig. 2.

In conclusion, we have shown that epitaxially grown silicon nanowires of radius greater than 20 nm prefer the <111> direction whereas wires with radii < 10 nm are mostly <110> oriented. This change of the growth direction has been explained by considering the free energy of the silicon nanowire. A quantitative estimate provides a crossover radius of 10 nm, in good agreement with experimental data.

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Piezoelectric ZnO nanowire arrays

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To realize the potential applications of ZnO-based 1D nanoscale materials in optoelectronic switches, high-efficiency photonic devices and near-UV lasers [1], it is of importance to fabricate high quality single-crystalline ZnO nanowires with defined orientation and position. A straightforward approach involves patterned metal nanoparticles which are used as catalyst templates for the subsequent guided vapor-liquid-solid growth of ZnO nanowires. A number of methods have been used to obtain patterned metal catalysts, including electron beam and imprint lithography, as well as direct deposition through masks such as grids and microsphere monolayers. We applied a new template method for the growth of well-ordered ZnO nanowires on conducting GaN layers [2]. The catalytic Au templates were produced by thermal evaporation using a novel type of metal nanotube membrane as shadow mask. This is an inexpensive and generic approach and it is in principle applicable for the fabrication of any semiconductor nanowire array on corresponding lattice-matched substrates.

The fabrication process involves three main steps. First, membranes comprising a hexagonal array of Au nanotubes were electrochemically fabricated by replicating the master structure of nanoporous anodic aluminum oxide (AAO) [3]. While the alignment and pattern of the nanotubes mirror the initial monodomain pore array of the AAO template, the nanotube inner diameter is adjustable. In our experiments, the membranes used have tube diameters of $\sim 60-130$ nm.

Second, in order to create the catalytic Au template, the membrane was transferred onto the substrate (GaN epilayers) surface via homogeneous contact. After thermal evaporation of Au through the nanotubes, the membrane was removed, leaving a hexagonal array of Au nanodisks. Figure 1 shows the deposited Au disk arrays, together with a leftover of the membrane piece. For the 130 nm nanotube membrane, the printed Au nanodisks have a mean diameter of 100 nm and inter-disk separation of 500 nm.

Finally, ZnO nanowires were grown via a vapor transport and deposition method on GaN(0001)/Si(111) layers using the Au nanodisks as catalyst. Figure 2 shows the obtained ZnO nanowire arrays with the same periodic order of the initial gold pattern. The nanowires are oriented perpendicular to the substrate surface with hexagonal ends, indicating their main axis is oriented along [0001]. The interwire distance is constant at 500 nm, as predefined by the tube-to-tube distance of the shadow mask. The average height of the wires after a growth time of 20 min is 1.5 μm. The diameters, however, are strongly dependent on the nanotube membranes used. The wire diameters are $87-103$ nm in Fig. 2a-b while $30-70$ nm in Fig. 2c, corresponding to the nanotube inner diameters of 130 and 60 nm, respectively. Furthermore, the ZnO wire size also depends on the growth conditions (temperature and time). For example, pillar ar-

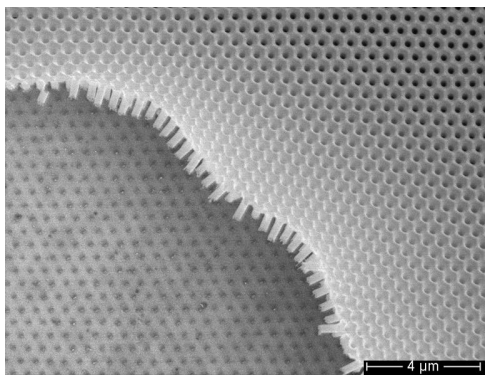

Fig. 1: SEM image of substrate surface after thermal deposition of Au using the nanotube membrane as shadow mask. Part of the membrane was removed (left), exposing the Au nanodisk array.

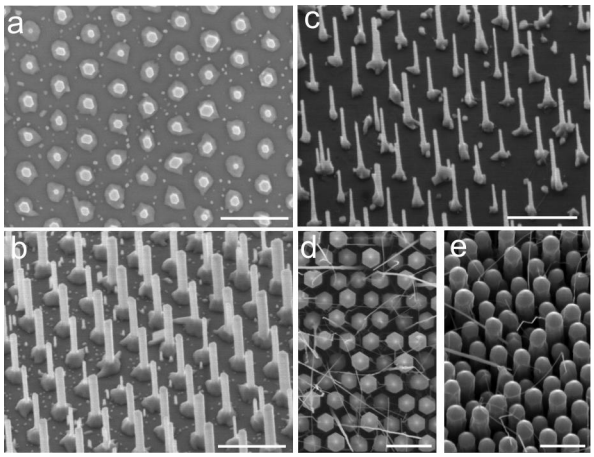

Fig. 2: SEM images of ZnO nano/sub-micron wire arrays. (a,d) Topview. (b,c,e) Inclined view. The nanotube inner diameters of the membrane are 60 nm for (c) while 130 nm for the rest. Note that the growth conditions for (a-c) differ slightly from those for (d-e) in temperature and synthesis time. Scale bars: 1 μm.

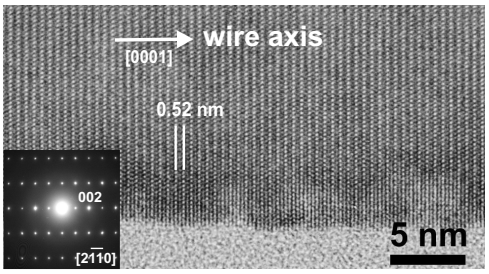

Fig. 3: HRTEM image of a segment of a ZnO nanowire. (Inset) electron diffraction pattern.

rays (see Fig. 2d-e) with pillar diameters in the 200 nm range were obtained using the same mask as that in Fig. 2a-b, but at a lower temperature and for prolonged times of growth.

The crystallographic orientation of the nanowires was determined using TEM (Fig. 3). Electron diffraction confirms that the nanowires are wurtzite single crystals with the main axis along [0001]. From the lattice image, the distance between the parallel planes along the wire axis corresponds to a *d*-spacing of the (0001)-planes of 0.52 nm.

For potential applications of our ZnO nanowires as piezoelectric transducers, their piezoelectric properties were investigated using piezoresponse force microscopy (PFM). The measurements were conducted on ≈150 nm diameter free-standing individual wires by

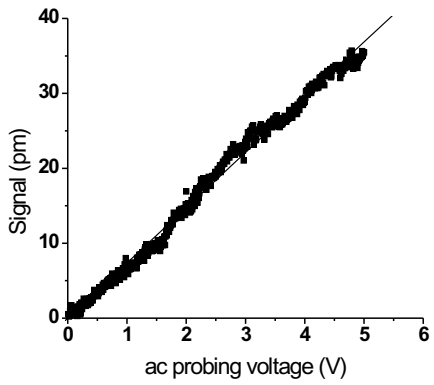

Fig. 4: Piezoelectric characterization of an individual ZnO nanowire with a probing frequency of 33.25 kHz. From the slope of the linear fitting, the piezoelectric coefficient d_{33} was determined to be 7.34±0.3 pm/V.

sweeping the applied ac voltage in the desired range (see Fig. 4). The highly-doped GaN layer provides the bottom electrode whereas the conducting tip acts as the top electrode. The effective piezoelectric coefficient d_{33} averaged over all measurements is 7.5±0.6 pm/V, as is reasonable compared with the bulk values of (0001) ZnO (9.93 pm/V) measured also using PFM. The frequency spectra of the PFM signal (not shown here) is relatively flat over the measurement range of 0−230 kHz. This is consistent with previous studies on PFM methods and the particular measurement geometry on the high aspect ratio nanowires that minimize the linear electrostatic contribution of the cantilever to the entire PFM signal [4].

Work on reducing the diameter of the nanowires as well as the distance between the nanowires is in progress.

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Ferroelectric Properties of P(VDF-co-TrFE) Copolymer Nanotubes

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Ferroelectric polymer nanotubes are promising nanostructures because they may combine outstanding chemical and mechanical stability with the functionality of ferro- and piezoelectric materials. Polyvinylidene fluoride (PVDF) has been widely studied for this purpose [1]. Recently, nanotubes or nanowires made of PVDF have been investigated [2]. However, from the melt or diluted solutions PVDF usually crystallizes in the non-polar α-phase (phase II). A conversion into the macroscopically polar β-phase (phase I), which shows ferroelectric properties, is only possible by stretching at elevated temperatures. Obviously, this procedure is not easily applicable to nanostructures. This drawback can be overcome by integrating trifluoroethylene (TrFE) units in the backbone of PVDF [3, 4]. The resulting statistical copolymer P(VDF-co-TrFE) avidly crystallizes in a polar phase similar to the ferroelectric β-phase of PVDF homopolymer. If the statistical copolymer contains VDF and TrFE at a ratio of 75:25 (Fig. 1 shows a corresponding unit cell), the highest spontaneous polarization can be achieved.

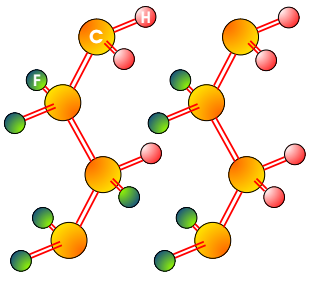

Fig. 1: Schematic of a unit cell of P(VDF-co-TrFE) containing VDF and TrFE at a ratio of 75:25. [3]

Nanotubes and nanowires with uniform size made from P(VDF-co-TrFE) copolymer with diameters D_p ranging from 25 nm to 1 μm have been fabricated on a large scale by wetting of ordered porous templates (such as macroporous silicon or porous alumina) and subsequent cooling at 1 K per minute under argon atmosphere. For this purpose we have adapted an approach investigated particularly by C. R. Martin, which involves the use of porous templates as molds [5]. Released tubes and rods with dimensions depending on those of the template pores can be obtained by selectively removing the template. As evidenced by scanning electron microscopy (SEM) (Fig. 2), these one-dimensional nanoobjects have smooth walls and high aspect ratios (up to 200).

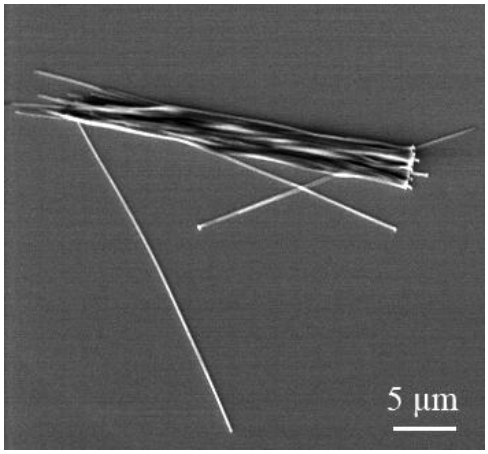

Fig. 2: SEM image of P(VDF-co-TrFE) nanotubes with a diameter of 180 nm and a length of 100 μm.

XRD measurements in the reflection mode were performed on P(VDF-co-TrFE) copolymer nanotube and nanowire arrays which were embedded within the alumina templates and thus perfectly aligned. The samples were placed in the X-ray diffractometer in such a way that the template surface with the pore openings was arranged perpendicular with respect to the plane defined by the incident and the scattered X-ray beams. The $\theta/2\theta$ scans (Fig. 3a) revealed that the P(VDF-co-TrFE) copolymer nanotubes with both D_p = 60 nm and 400 nm contain crystalline domains con-

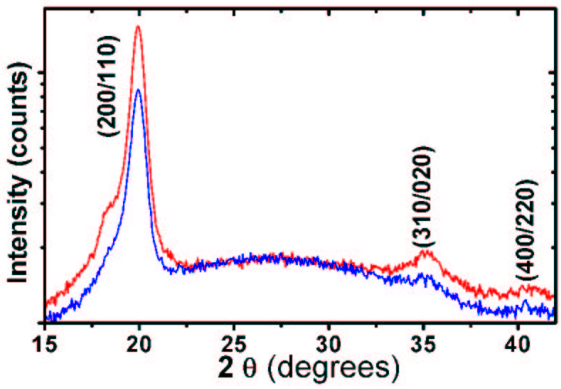

Fig. 3: X-ray patterns of P(VDF-co-TrFE) nanotubes. Red: D_p = 60 nm; blue: D_p = 400 nm.

sisting of the analogue to the polar β-phase of PVDF.

In order to probe local ferroelectric behaviour, individual P(VDF-co-TrFE) nanotubes were characterized by AFM (Atomic Force Microscopy) imaging and PFM (Piezoresponse Force Microscopy) performed in contact mode with a conductive cantilever (force constant about 0.25 N/m). We deposited the nanotubes on gold coated mica by dip coating from an ethanolic suspension. Fig. 4a shows the topography of an individual P(VDF-co-TrFE) nanotube with a diameter of 400 nm. Local (remnant) piezoelectric hysteresis loops were measured by successive application of rectangular voltage pulses. The PFM signal was measured after the voltage had been switched back to zero after several poling procedures. Sub-μm domains along the long axes of the tubes could be well resolved in the PFM images (amplitude and phase shown in Fig. 4b and Fig. 4c respectively) in which hysteresis loops were recorded. Fig. 4d shows the virgin (in blue) and the hysteresis loops (in red) before and after poling, respectively. The fact that the hysterisis loop is not closed is related to relaxation phenomena.

Further work will focus on the fabrication of extended ordered arrays of ferroelectric polymer nanotubes. Such structures consist of perfectly aligned free-standing P(VDF-co-TrFE) nanotubes with their open end connected to a P(VDF-co-TrFE) membrane and

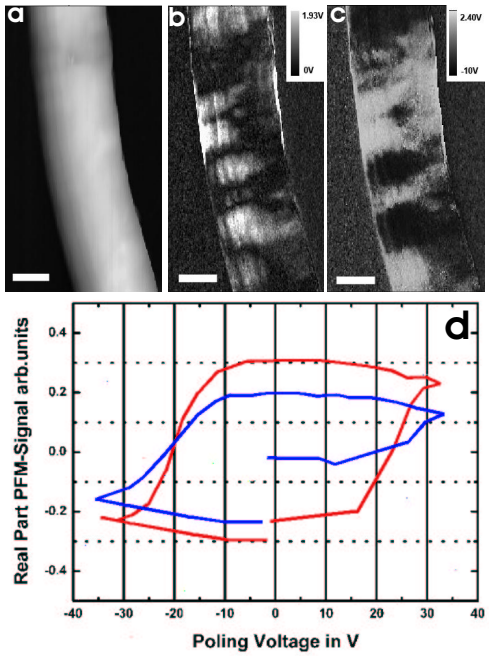

Fig. 4: AFM and PFM characterization of an individual P(VDF-co-TrFE) nanotube with a 400 nm pore diameter. (a) AFM topography image; (b) PFM amplitude image; (c) PFM phase image; (d) ferroelectric hysteresis loop before (blue) and after (red) poling. The scale bar represents 200 nm in all the images.

the capped end pointing upward. Especially with electrodes integrated on selected areas such as inner and/or outer tube walls they may be used for building blocks in miniaturized ultrasonic sensors and piezoelectric micro-electromechanical systems (MEMS).

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Ultrathin strained silicon layers on insulators (sSOI) by wafer bonding

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For more than 30 years, MOS device technologies have improved at a steady rate. A large part of the success of the MOS transistor is due to the fact that it can be scaled to nanoscopic dimensions (< 100 nm), which results in higher performance without getting to its physical limits. Transistor delay times have decreased by more than 30% per technology generation resulting in the doubling of microprocessor performance every two years. This behavior is known as Moore's law. In order to maintain this pace of improvement even in the nanotechnology regime it is not sufficient just to shrink devices, but it is also required to integrate and engineer novel materials and novel device structures. Among these novel materials we find on the ITRS (international technology road map for semiconductors) strained silicon instead of unstrained silicon in the device channel and silicon on insulator (SOI) materials, i.e. materials where the device channel is separated from the bulk of the Si wafer by a buried oxide (BOX). While strain alters the band structure of the Si in a way that enhances carrier mobility [1], the SOI structure provides for reduced leakage currents through the body of the wafer and reduced capacitance [4]. In combination, i.e. in strained silicon on insulator (sSOI) a remarkable performance enhancement is obtained [2] without any shrinkage of the transistor. This sSOI material is targeted for future CMOS processing and consists of a tensile strained silicon layer of a thickness between $\approx$ 10 nm and $\approx$ 50 nm on a buried oxide of 100-200 nm thickness on a Si (001) wafer. To obtain such sSOI wafers, the technique of direct wafer bonding has nowadays been established [2]. It relies on the fact that two solids (here wafers) with sufficiently flat, smooth (root mean square roughness < 1 nm) and clean surfaces adhere to each another when brought in close contact in air at room temperature. The attraction between the two wafers is primarily mediated through vander-Waals forces or hydrogen bridge bonds. With a suitable additional annealing step the bonding energy of the interface may reach up to the cohesive energy of the materials.
Figure 1 shows our bonding procedure to obtain sSOI. We have one oxidized Si (001)

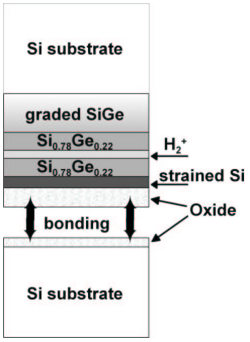

Fig. 1: Schematic of wafer bonding proceedure to create sSOI: the top wafer is the device wafer, the bottom wafer the handle wafer.

(called handle wafer) and a second one (called device wafer) containing the strained silicon as the top layer, which is may be covered by a thin oxide layer, too. A strained silicon layer with the nanoscopic thicknesses required, at a high degree of strain and layer thickness uniformity throughout the entire wafer, is obtained by epitaxial Si deposition (sufficiently low thickness to avoid plastic relaxation) on a relaxed SiGe buffer layer, which has a larger lattice constant (depending on the SiGe-alloy composition between 0 and 4.2%) and which forms a so called 'virtual substrate' together with the Si (001) wafer. The relaxed SiGe-buffer layer is obtained by epitaxial deposition too, however, at layer thicknesses large enough to encourage close to 100% plastic relaxation. For the removal of the virtual sub-

strate after bonding, hydrogen implantation induced splitting and layer transfer is used. A strong bond between the two wafers is formed by a 12 h anneal at a temperature of 200 °C. Once the bond has formed, a second higher temperature anneal (e.g. 2 h at 500 °C) induces the splitting process that separates the bonded wafer pair parallel to the surface at a thickness ≈ 700 nm below the surface. After splitting, the root mean square (RMS) surface roughness is homogeneous over the entire wafer and is of the order of 15 nm. The total thickness variation of the transferred layer over the entire wafer is of the same order. The 700 nm thick SiGe-layer, remaining on the bonded wafer is removed by selective etching at room temperature using buffered HF (5%):H_2O_2 (31%):CH_3COOH (100%) at a mixture of 1:2:3. Figure 2 shows cross section transmission electron micrographs (TEM) of the as grown wafer showing the strained silicon layer atop the virtual substrate (C), and the transferred strained silicon layer atop the buried oxide after selective etching of the virtual substrate at different magnifications (A,B). The amorphous layer atop the strained Si is glue from the TEM sample preparation. Selective defect etching (modified Schimmel

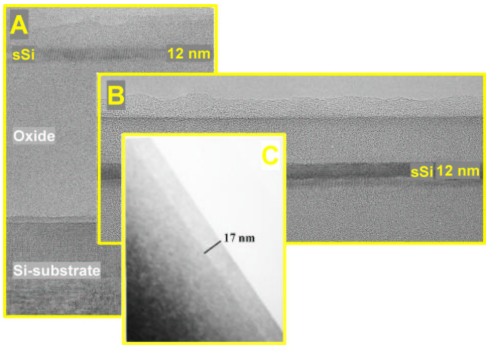

Fig. 2: Cross section transmission electron micrographs of the as deposited strained silicon on the virtual substrate ($Si_{0.72}Ge_{0.28}$) (C) and the strained silicon (sSi) after selective etch removal of the virtual substrate on the buried oxide (A, B), while B shows a higher magnification of the layer.

etch [3]) indicates the threading dislocation density of the transferred strained Si of about $10^5 cm^{-2}$, which is a number comparable to that of the relaxed SiGe-layer. This means no additional dislocations are generated by the bonding procedure. Figure 3 shows a micro-Raman analysis of the transferred layer. Unstrained silicon has a characteristic Raman peak at 520 cm^{-1} (for reference inserted blue in Figure 3), tensile strained silicon has a peak at lower values. SiGe-alloy layers have peaks depending on their composition: Si-Si reside near 500 cm^{-1}, Si-Ge peaks reside near 400 cm^{-1} and Ge-Ge peaks reside near 300 cm^{-1} [5]. In our Raman spectrum no indication of SiGe has been found, showing that the selective etching successfully removed the entire remaining virtual substrate after the layer transfer process. A strained silicon peak (red) at 514 cm^{-1} indicates strain of the order of 0.8% [5].

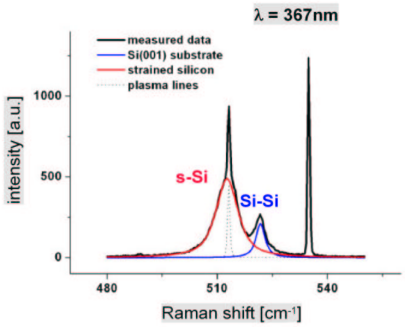

Fig. 3: UV micro-Raman analysis of the transferred strained silicon layer on buried oxide layer (red). For reference an unstrained silicon characteristic Raman peak (blue) is inserted. The two sharp peaks are plasma lines from the laser.

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Polarization imprint in ordered arrays of epitaxial $BaTiO_3$ and polycrystalline $SrBi_2Ta_2O_9$ ferroelectric nanostructures

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Future non-volatile, high-density ferroelectric random access memories [1] in the Gbit range require studies of size effects in arrays of ferroelectric nanostructures with individual structure sizes below 100 nm. Possible size effects are the loss of ferroelectricity due either to depolarization fields or to the loss of long range cooperative driving forces, the suppression of polarization switching due to either domain pinning or to the so-called imprint effect corresponding to a preferential polarization state and an asymmetry in the polarization hysteresis loop [2]. The recent development of local piezoresponse measurement techniques based on scanning force microscopy (SFM) has enabled nanoscale investigations of polarization switching in individual ferroelectric capacitors [3]. A self-assembled monolayer of monodisperse latex spheres arranged in a close-packed hexagonal array was used as a lift-off-type deposition mask [4] for pulsed laser deposition (PLD) of $BaTiO_3$ (BTO) or $SrBi_2Ta_2O_9$ (SBT) nanostructures. These two materials are prototypes for perovskite and bismuth-layered perovskite ferroeletric materials, respectively. The latex monolayer was prepared on a conductive (100)-oriented Nb-doped $SrTiO_3$ single crystal substrate by a spin-coating process, using commercial monodisperse polystyrene latex spheres of 1 µm or 0.5 µm diameter. Figure 1 is an SEM image of a monolayer. The bright regions correspond to stacking faults. PLD of $BaTiO_3$ and SBT from corresponding ceramic targets was performed in oxygen at $1{\cdot}10^{-6}$ Torr at RT, using a KrF excimer laser with an energy of 400 mJ at a repetition rate of 1 Hz. After deposition the polystyrene latex spheres were lifted off in methylene chloride. Subsequently the as-deposited nanostructures were annealed for 1 h in air at 650°C (BTO) and 950°C (SBT), respectively.

Figure 2 shows an SEM image of part of an SBT array obtained using a 1 µm latex sphere mask, and Fig. 3 is a corresponding AFM topography image. The $BaTiO_3$ nanostructures are epitaxially crystallized, as can be judged

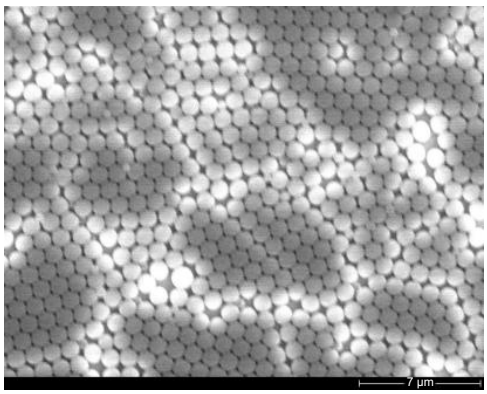

Fig. 1: SEM image of a monolayer of latex spheres.

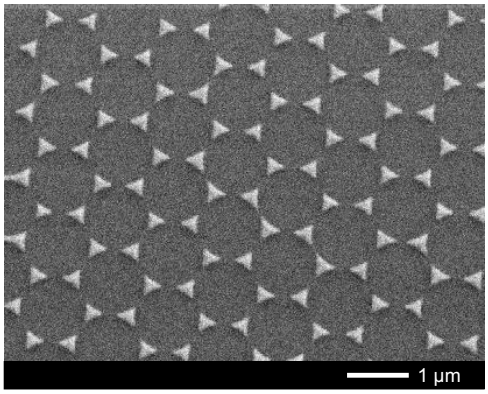

Fig. 2: SEM image of part of a $SrBi_2Ta_2O_9$ array.

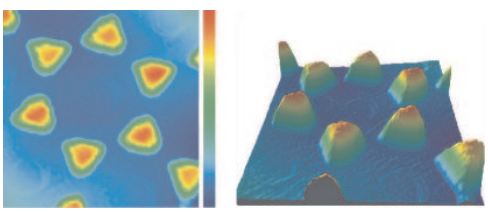

Fig. 3: AFM topography images (2 µm x 2 µm) of part of a $SrBi_2Ta_2O_9$ array in 2D and 3D view.

from TEM images and electron diffraction patterns of cross sections prepared by focussed ion beam thinning (Fig. 4), whereas the SBT nanostructures are polycrystalline and involve a bottom layer of a different phase formed due to reactive interdiffusion (Fig. 5). For more details, see refs. [5-7]. Hysteresis loops (Fig. 6) taken by piezo-response scanning force microscopy (PFM) show a thickness-dependent imprint effect of both the BTO and the SBT nanostructures. Whereas the hysteresis curve (a) of a 44 nm high BTO nanostructure (fabricated from a 1 µm sphere mask) is more or less symmetric, the loops (b) and (c) of two only 26 nm high nanostructures (fabricated using a 0.5 µm sphere mask) are shifted towards the negative d_{33} axis, i.e. are imprinted. This effect is independent of the lateral size. (Curve (c) was recorded from a laterally extended structure formed at a stacking fault of the sphere mask.)

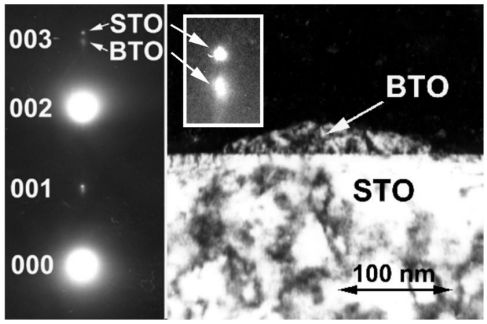

Fig. 4: Cross section (001) dark-field TEM image of a BTO nanostructure. Left, and inset: Electron diffraction pattern demonstrating the epitaxy.

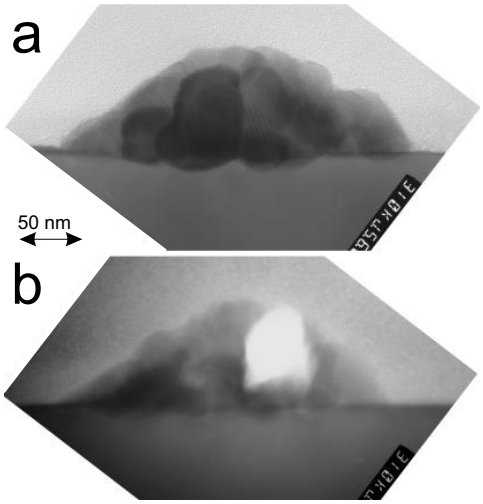

Fig. 5: Cross section TEM images of an SBT nanostructure. (a) Bright-field image, (b) $(022)_{SBT}$ dark-field image.

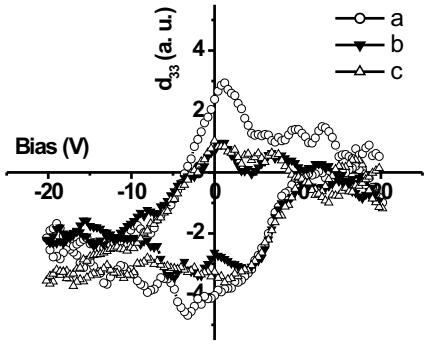

Fig. 6: PFM hysteresis loops obtained from BTO nanostructures of different size. d_{33} is the effective piezoelectric coefficient along the vertical of the nanostructure. For curves a, b, and c, see the text.

According to our model [2,6], the imprint effect in the $BaTiO_3$ nanostructures results from a domain locking near the ferroelectric-electrode interface, so that the nanostructures are composed of switchable regions and of non-switchable ones, the latter being close to the interface with the substrate. High-resolution TEM investigations are in progress to evaluate any structural differences between the volume of the nanostructures and their near-interface regions.

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Microstructure, nanochemistry, and electronic properties of interfaces in 4H-SiC MOS devices

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An important advantage of the semiconductor silicon carbide compared to other wide band gap semiconductors is its native silicon dioxide which is readily formed in an oxidizing ambient at elevated temperatures. This property enables the fabrication of metal-oxide-SiC field effect transistors (MOSFET) similar to silicon devices. The most favorable SiC polytype for such a device is 4H because of its large bandgap (3.26 eV) and a bulk electron mobility of approximately 800 cm^2/Vs. However, the great potential of the n-channel 4H-SiC MOSFET devices has thus far been hampered by an unacceptably low inversion channel mobility which is attributed to the high density of interface states near the SiC conduction band edge. Large portion of these interface states of unknown origin are so called slow states or near-interface traps (NITs) that are not revealed by conventional capacitance-voltage (C-V) analysis at room temperature.

The purpose of this work is to look for a possible correlation between microscopic defects at the SiO_2/SiC interface and the near-interface traps close to the 4H-SiC conduction band edge. We combine high resolution electron microscopy (HREM) with nanochemical analyses via electron energy- loss spectroscopy (EELS), especially energy-loss near-edge structures (ELNES), and energy filtered electron microscopy (EFTEM), and we use C-V analysis and thermal dielectric relaxation current (TDRC) measurements – where the majority carriers are thermally emitted from interface traps – to investigate the electronic peculiarities of SiO_2/SiC interfaces and MOS gate stacks typical for n-channel MOSFETs [1]. The samples were derived from commercially grown highly doped 8 degrees off-axis n-type 4H-SiC. Fig. 1(a) shows C-V curves from simultaneous Hi-Lo measurements using

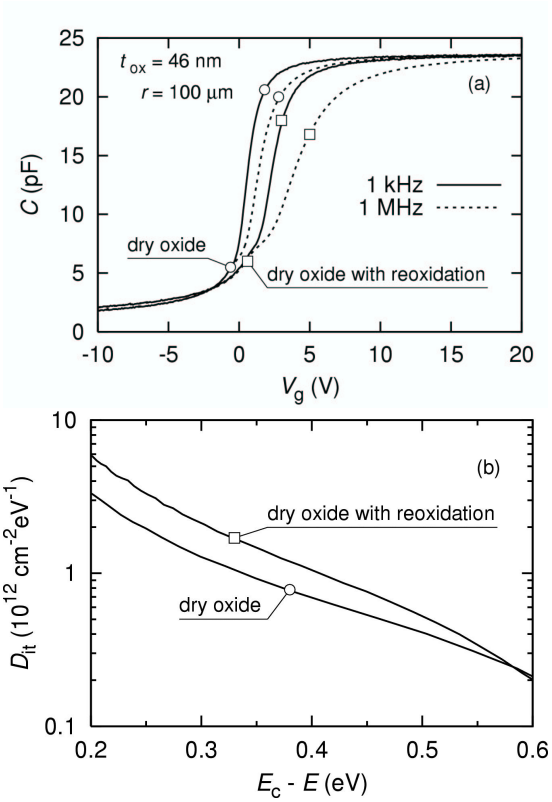

Fig. 1: (a) C-V measurements on n-type 4H-SiC MOS capacitors. (b) Interface state density extracted from the data in (a).

1 MHz and 1 kHz probe signals from n-type 4H-SiC MOS capacitors for a sample with dry oxide and for a sample receiving reoxidation anneal. The density of these interface states estimated from the frequency dispersion is plotted in Fig. 2(b). The re- oxidation results in a slight increase of the interface state density near the conduction band. The C-V curves for the re-oxidized samples are also shifted to the right as compared to dry oxide, which is either due to an increase in the fixed negative charge or an increase in the density of deep electron traps that once filled do not thermally emit their captured electrons at room temperature within the time frame of the measurement.

The physical origin of these electron trapping interface states remains unclear up to

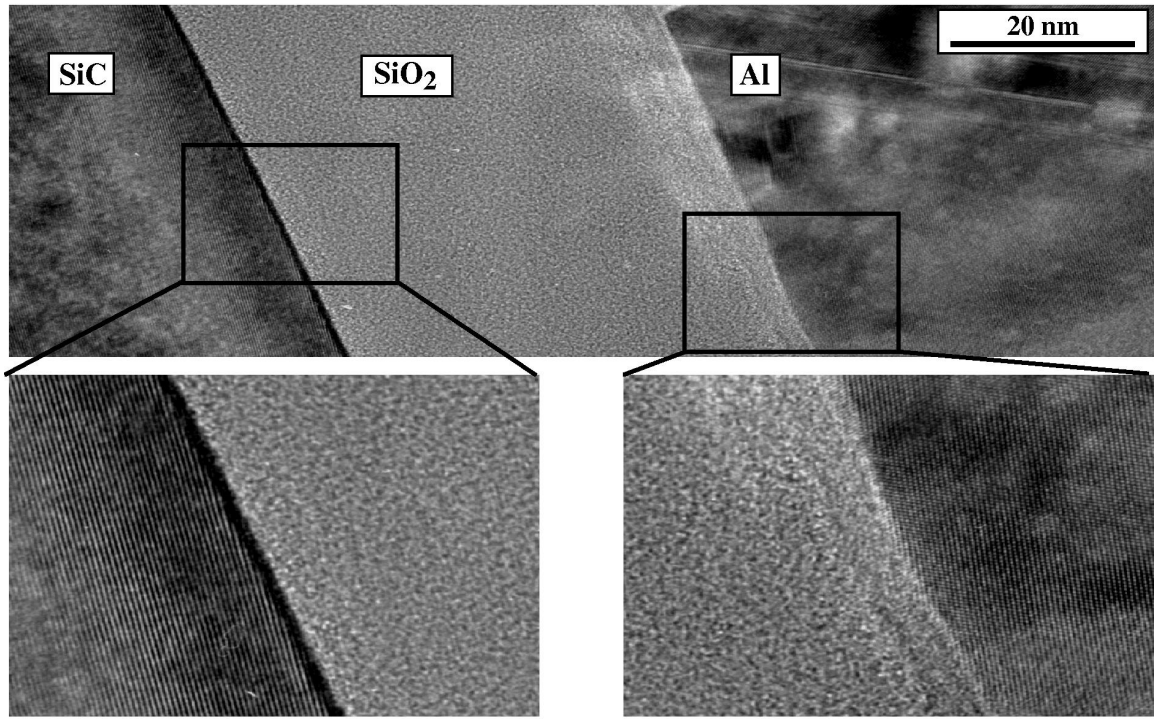

Fig. 2: HREM images of the SiO_2/Al layer system on a 4H-SiC wafer (below; enlargements of the indicated areas)

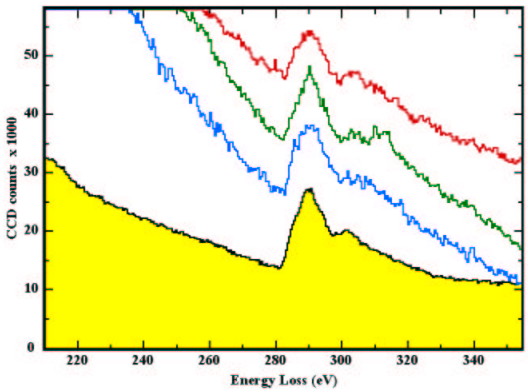

Fig. 3: Electron energy loss spectra of the carbon-K edge recorded at different places along the SiO_2/SiC interface line (cf. Fig. 2, left). The lowest spectrum amounts to the SiC standard.

now, but the most discussed idea is these traps being associated with electrical active carbon clusters [2]. However, our HREM observations (cf. Fig. 2) reveal that the SiO_2/SiC interface is abrupt, and there are no graphitic structures detected in the amorphous silicon dioxide or near both interfaces; also stacking faults or other planar defects were not observed in the SiC near the interface. Fig. 3 shows the carbon-K energy loss edge spectra recorded just at the SiO_2/SiC interface. The spectra give no hints to any π-bondings of graphitic or other double bonded carbon and agree well with the carbon signals in standard SiC, indicating that all the carbon detected near the interface exists in the form of SiC bonded carbon.

Altogether, the investigations show that the NITs cannot result from carbon clusters or graphitic regions in the SiO_2/SiC interlayer or within the SiO_2. Therefore, they must be caused presumably by intrinsic oxygen defects extending from the interface into the oxide layer as proposed in [3].

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The morphology of DNA-conjugated gold nanoparticles

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Nowadays there are many possibilities to create nanostructures with different shapes and from different materials. Nevertheless, the generation of building blocks with specific architectures and properties is still a challenge. A promising strategy is to use artificial DNA as linker for nanostructures. The first and basic papers in the field of DNA-linked nanoparticles have been published in 1996 [1, 2]. The main condition of the DNA-mediated linking of particles with single-stranded DNA (ssDNA) is the specific recognition of the two ssDNA strands. These regocnition leads to generation of particle networks (Fig. 1).

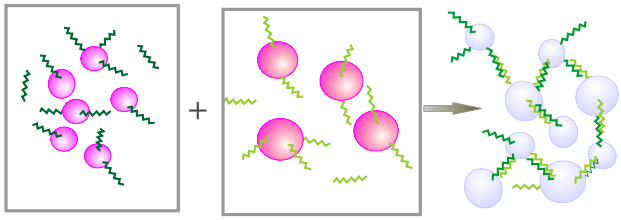

Fig. 1: Schematic description of the stategy to generate particle networks with single-stranded DNA.There are two fractions of particles. Each is functionalized with single-stranded DNA (ssDNA). These ssDNA strands are complementary to each other and hybridize, when particles are mixed. As a consequence particle networks form.

Up to now however, there is little known about interactions between the DNA and the particle surface, the role of the necessary cations and related to this, the distinction between DNA-mediated and unspecific generation of particle aggregates. Further, there is little known about the reversibility of particle linking. Therefore we linked gold nanoparticles via double-stranded DNA (dsDNA). The use of dsDNA, which is mechanically stiffer than ssDNA, allows the investigation of parameters influencing the particle linking as well as interactions between DNA and the particles. This is a prerequisite for the development of robust linker structures and reliable protocols for mass production. In the following,various aspects of the morphologies of DNA-linked gold nanoparticles will be discussed.

We coupled gold nanoparticles with diameters of 20, 50 and 80 nm. Gold nanoparticles are a well-established test system to investigate DNA-linking. For the investigation of the DNA-mediated coupling of particles, mainly absorption spectroscopy was employed. This allowed a non-invasive investigation of the aggregates in solution. Since DNA-linker change the distance between linked particles, the plasmon resonance frequency changes, too. The successful coupling of particles leads to a redshift and broadening of the peak. To what extend the redshift occurs, depends on the particle diameter. The peak broadening is a result of the generated aggregate sizes. The particles were coupled with 24mer double stranded DNA, which has an approximate length of 8 nm. The DNA is functionalized with a thiol group at both ends. The particles were mixed with DNA. For DNA-linkage of gold particles a salty environment is necessary. The method of dsDNA-linking of nanoparticles is schematically illustrated in Fig. 2.

Fig. 3 shows an example for the absorption spectrum of dispersed and dsDNA-linked 20 nm gold particles. In case of dispersed particles the wavelength is 525 nm and shifts to 555 nm in case of linked particles. The generated aggregates were additionally investigated

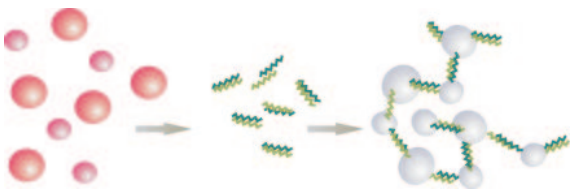

Fig. 2: Schematic description of the applied stategy to generate particle networks. The particles were mixed with the double-stranded DNA. The dsDNA binds covalently via the thiol group on the surface of the gold nanoparticles. This leads to the linking of particles to a particle network.

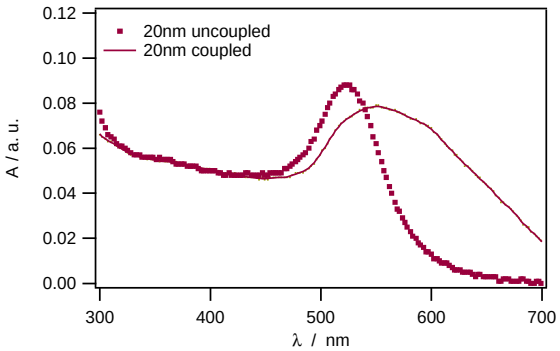

Fig. 3: Absorption spectra of 20 nm gold nanoparticles. The dispersed particles have an absorption maximum at 525 nm (dotted line). The absorption maximum of dsDNA-linked particles is redshifted to 555 nm (solid line).

by transmission electron microscopy (TEM). The analysis yields that the number of particles in an aggregate decreases as the particle size increases. In case of 20 nm gold particles most of the aggregates consist of 30 to 50 particles. In case of the 50 nm particles half of the aggregates contain 10 to 29 particles and the aggregates of 80 nm gold particles on average contain 17 particles (Fig. 4).

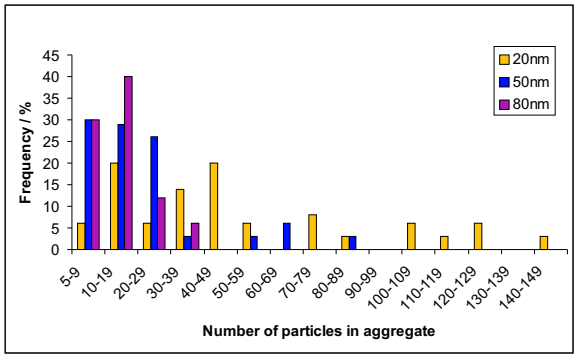

Fig. 4: Distribution of number of particles in an aggregate of dsDNA-linked gold nanoparticles. With increasing particle size, the number of particles in an aggregat decreases.

Besides a coupling of particles with the same diameter, the coupling of bigger and smaller ones is also possible. Fig. 5a illustrates the color change of suspension containing dsDNA-linked gold nanoparticles with different particle diameters and different ratios compared with the dispersed particles. The color of the suspension of the dispersed gold

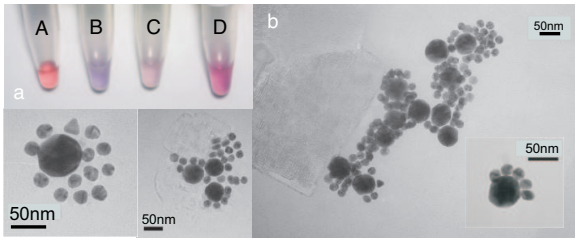

Fig. 5: DNA-linked gold nanoparticles with different diameters. **(a)** Color change of linked gold nanoparticles **A:** uncoupled colloids, **B:** 20 nm and 80 nm, **C:** 50 nm and 80 nm **D:** 20 nm and 80 nm (different mixing ratio). **(b)** TEM-image of DNA-linked 20 nm and 50 nm gold nanoparticles.

nanoparticles is red and changes in case of coupling to pink or purple. The TEM-images of gold nanoparticles are represented in Fig 5b. In this case 20 nm and 50 nm gold nanoparticles were DNA-linked. Since there is no limiting factor to what extend and where the particles link to each other, the resulting aggregates may be ordered as well as disordered. The use of double-stranded DNA as linker allows in solutions with different sizes of particels additionally a coupling of same-size particles.

Besides the results presented here, it is also necessary to pay attention on the role of the added salt and interactions between DNA and particles [3]. Furthermore we worked on the evidence and reversibility of DNA-linking as well as the different behavior of ssDNA and dsDNA. In future we plan to connect different nanostructures to each other via a much more robust and selective DNA linker, whose structure is the result of the knowledge we earned.

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Many-body effects in nanospintronics devices

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The nanostructure physics has realized experiments for important quantum effects. For example, in a quantum dot (QD), the strong Coulomb interaction suppresses *real* charge fluctuations and thus the QD with odd number of electrons behaves as a local spin 1/2. Then the Kondo effect [1], which is a basic concept in the field of magnetism as the Ruderman-Kittel-Kasuya-Yosida (RKKY) interaction, occurs. Recent fabrication technique of magnetic nanostructures evolved the active research field "spintronics" [2]. The nanospintronics device possesses basic interests since it provides a way to control the spin degrees of freedom and can realize novel many-body states. We analyzed the possibility of the Kondo effect in a new geometry, a QD coupled to two ferromagnetic leads [3]. We also theoretically investigated the RKKY interaction in a semiconducting nanostructure [4], which has not been addressed though the RKKY interaction is a well established mechanism for magnetic nanostructures [5]. Discussed many-body effects were very recently demonstrated experimentally [6, 7].

The Kondo effect in QD reveals itself as the zero-bias anomaly of nonlinear differential conductance. If two lead electrodes are spin-polarized and their magnetizations are parallel, the local spin inside the QD tends to align to the direction of magnetizations because of spin dependent virtual (*quantum*) charge fluctuations [3]. Figure 1(1) shows a typical situation of QD, where a single level ϵ_0 inside QD laying below the left (right) electrode Fermi energy. In such a case, an electron with majority spin inside QD can gain the kinetic energy, because it can fluctuate between leads and the QD more than an electron with minority spin. Thus the majority spin is stabilized just in the same case as the kinetic exchange interaction [8] and the spin degeneracy of QD level is lifted by $\Delta\epsilon$. The perturbative scaling analysis indicates the spin splitting is proportional to the average value of spin polarization factors in left and right leads. The spin splitting quenches low energy spin flip scattering processes when the bias voltage V is small, $|eV| < |\Delta\epsilon|$, and effectively splits the zero-bias anomaly as schematically shown in Fig. 1(2).

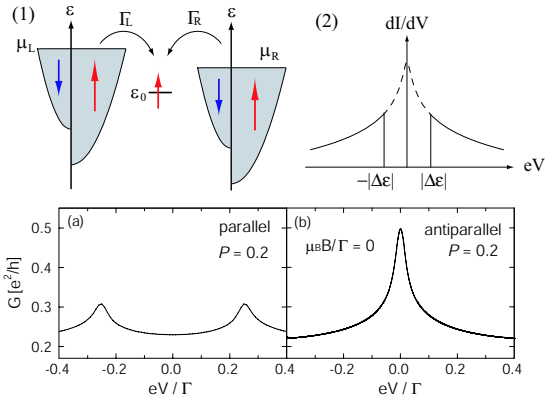

Fig. 1: (1) A quantum dot coupled to two ferromagnetic leads. (2) A schematic plot of the zero-bias anomaly (dashed line) and split zero-bias anomaly (solid line). The differential conductance as a function of the applied bias voltage V for parallel (a) and antiparallel (b) alignments.

Figures 1 (a) and (b) show the differential conductance as a function of the bias voltage. For parallel alignment, a splitting of the peak is obtained (a). In anti-parallel alignment, the zero-bias anomaly is recovered (b), because the average value of spin polarization factor is zero. Recently, the Kondo effect is observed in single-molecule transistors. Especially, a C_{60}

molecule coupled with ferromagnetic leads [6] confirmed our prediction.

The RKKY interaction itself has been known from the 1950s [9]. Motivated by the progress of nonofabrication technique in 2-dimensional electron gas in a GaAs/AlGaAs heterostructure, we analyzed the possibility of the RKKY interaction in an Aharonov-Bohm (AB) ring embedded with a QD in each arm [inset of Fig. 2 (a)]. Because of the long Fermi wave length and weak decay of the RKKY interaction in lower dimension, it could be possible to observe RKKY interaction. Actually, the RKKY interaction has been demonstrated for two QDs coupled indirectly thorough an open QD [7].

For the ring geometry, an electron wave function splits into two ways and acquires the phase during the propagation for each path, and then interferes with itself. When the magnetic field is applied, an additional phase ϕ related with the magnetic flux is counted when an electron goes round the ring. Then the interference effect and consequently the conductance depends on the flux (AB effect). The flux also can control the RKKY interaction, because the particle-hole excitations enclose the flux [inset of Fig. 2 (a)].

The flux dependent RKKY interaction can entangle two QD spins and the spin state depends on the flux. Then it also affects the conductance. Panels (a) and (b) are the ϕ dependent conductance for ferromagnetic (F) and anti-ferromagnetic (AF) coupling cases. At zero or one flux, the maximum interaction is induced because of the constructive interference. For F coupling case, a triplet state is formed at low temperature and the conductance is enhanced by Kondo correlations. For AF coupling case, at low temperature, the singlet is formed and the conductance is suppressed. At half flux, the RKKY interaction is switched off by the destructive interference. Surprisingly the maximum is observed for both panels where, usually, we expect the suppression of the conductance. By combining above behaviors, for AF coupling case, the phase of AB oscillations is shifted by π, and for F coupling case, an additional maximum appears at half-integer values of the flux. Such characteristic behavior will be a clear evidence of the RKKY interaction in our system.

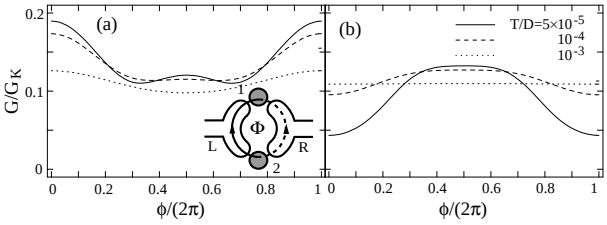

Fig. 2: Inset of panel (a): Aharonov-Bohm ring embedded with one QD (denoted by 1 or 2) in each arm and the flux dependent particle-hole excitation. Flux dependent conductance for (a) ferromagnetic and (b) antiferromagnetic coupling cases.

Overall, we investigated many-body effects in nanostructure devices which can be applied to the future nanospintronics devices.

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Ultrafast control of electronic motion in semiconductor quantum wells

A. Matos-Abiague and J. Berakdar

The time-dependent electric field in a subpicosecond unipolar electromagnetic pulse resembles one-half of an optical field cycle in an electromagnetic wave, therefore such pulses are referred to as "half-cycle" pulses (HCPs) [1]. For a freely propagating electromagnetic wave the time integral over the electric field $\mathbf{F}(t)$ vanishes. Therefore, an HCP is in fact a strongly asymmetric mono-cycle pulse (cf. Fig. 1 b) consisting of a very short strong half-cycle with a duration t_d followed by a much slower half-cycle of an opposite polarity and a much smaller amplitude (the tail of the HCP). Typical pulse amplitude asymmetry is 13:1 [1]. The interaction of HCPs with charge carriers is entirely different from that of continuous wave (cw) lasers: if the round-trip time of a confined electron is longer that the duration t_d of the HCP, the electron-HCP interaction can be viewed classically as an impulsive "kick" received by the charge carrier [2]. The amount of the kick (the momentum change) is given, in atomic units, by $\mathbf{\Delta p} = -\int_0^{t_d} \mathbf{F}(t)dt$. Quantum mechanically, subjecting an electron to an HCP results in a linear transformation of the momentum space wave function in the direction of the kick, $\tilde{\Psi}(\mathbf{p}) \rightarrow \tilde{\Psi}(\mathbf{p} + \mathbf{\Delta p})$. In the configuration space, applying an HCP phase-shifts the electron wave function as $\Psi(\mathbf{r}) \rightarrow \Psi(\mathbf{r})e^{-i\mathbf{\Delta p}\cdot\mathbf{r}}$. The weak tail of the HCP acts as an offset dc field that hardly affects the electron dynamics. From this scenario of the electron-HCP interaction one may expect that the position and the momentum of a given electronic distribution can be, to a certain degree, controlled and manipulated by applying a sequence of kicks with appropriate relative strengths, delays, and directions. Such HCP trains are feasible nowadays [3]. Recently, we have demonstrated [4] that using a train of HCPs allows a coherent control, on the *subpicosecond* scale, of the electronic motion in a $Al_xGa_{1-x}As$ based double quantum well. Such a possibility is desirable technologically, e.g., for the design of ultrafast switches or for the construction and control of quantum logic states (e.g., one associates 1 or 0 with the states where the electron is in the left or the right well, respectively (cf. Fig.1a)).

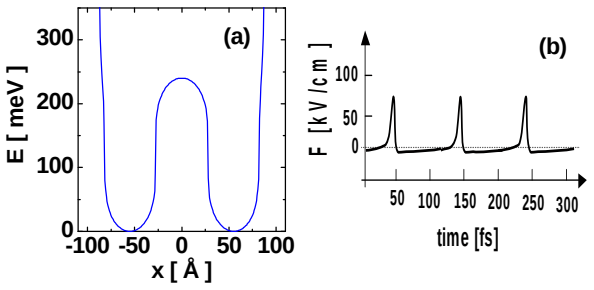

Fig. 1: (a) Electron confining potential: the central barrier hight is ~ 240 meV. The wells and barrier widths are ~ 50 Å and ~ 60 Å, respectively. (b) The electric field amplitude vs. time for a typical sequence of HCPs. .

Theoretically, one considers a conduction electron confined in a typical $Al_xGa_{1-x}As$ based double quantum well depicted in Fig. 1(a). Within the parabolic band and the effective mass approximations (the effective mass $m^* = 0.067m_0$ is assumed constant throughout the heterostructure), the Hamiltonian describing the system is

$$H(t) = H_0 + V_{conf} + V(x,t) \quad , \qquad (1)$$

where H_0 is the bare Hamiltonian, V_{conf} refers to the confinement potential (cf. Fig. 1 a) and $V(x,t)$ stands for the coupling of the electron to the pulses. As shown below a typical localization time is ~ 132 fs which is well below the typical time scale for the elastic scattering and electron-phonon interaction (several picoseconds) in high quality Ga(Al)As-GaAs heterostructures with typical electron concentrations [6]. Therefore, these effects are subsidiary for the localization process. The electron coupling to the sequence of (Gaussian-like shaped) HCPs is modelled by the time-

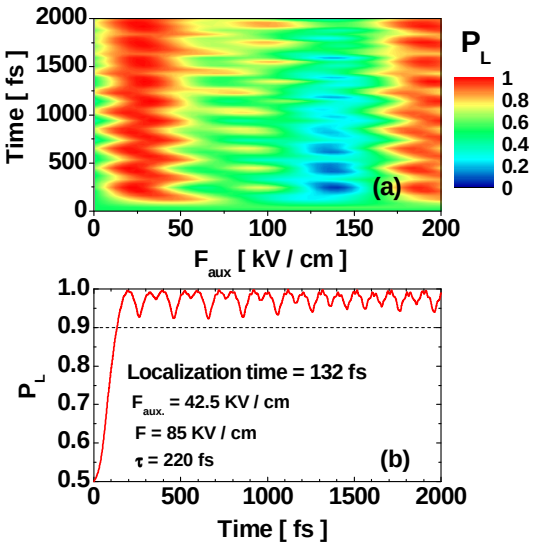

Fig. 2: (a) P_L vs. time and peak amplitude F_{aux} of the auxiliary pulse. Upon a time delay $\tau = 220$ fs after F_{aux} we apply a quasiperiodic train of HCPs with peak amplitudes $F_k = 85$ kV/cm (see text for details). (b) a cut in (a) at $F_{aux} = 42.5$ kV/cm .

dependent potential

$$V(x,t) = x \sum_{k=0}^{N-1} \left(F_k e^{-\frac{(t-t_0-kT)^2}{2\sigma^2}} \right) \quad , \qquad (2)$$

where σ characterizes the width of the pulses, t_0 corresponds to the time at which the first applied pulse is centered, T is the time between consecutive pulses, N is the number of applied pulses, and F_k is related to the amplitude of the k-th pulse. The time-dependent Schrödinger equation (1) is solved numerically using a fast-Fourier-transform based numerical method for the time propagation of the initial wave function. Having determined the time-dependent wave function $\Psi(x,t)$, we calculate the time-dependent probability [$P_L(t) = \int_{-\infty}^{0} \Psi^*(x,t)\Psi(x,t)dx$] and the time averaged probability [$\langle P_L \rangle_\tau = \frac{1}{\tau}\int_0^\tau P_L(t)dt$] of finding the electron in the left well. All the calculations were performed with $\sigma = 10$ fs, the first pulse was centered at $t_0 = 40$ fs, and the period was taken as $T = 100$ fs. Using an analytical two-level system (TLS) it has been possible to establish a guiding model to determine the parameters of the pulses that are appropriate for achieving a maximal, sustainable localization $P_L(t)$ of the electrons [4]. With the aid of the predictions delivered by the TLS model we identified the physical scenario and the pulses appropriate for an effecient localization: maximal localization is achieved if an auxiliary HCP of a strength F_{aux} is applied, followed, after an appropriate time delay τ, by a quasi periodic train of HCPs. In effect, the first pulse pushes the electron into the left well and the subsequent HCPs train keeps the particle localized in that well by "kicking" it back at the time when it starts tunnelling to the second well. This situation in illustrated by Fig.2 which shows full numerical results including the complete spectrum. Fig. 2 evidences that a strong localization of an initially delocalized state can be achieved in approximately hundred femtoseconds. This finding is in sharp contrast to the case when cw lasers are used as driving fields [5]; there it has not been possible to achieve such a strong localization and, in addition, the time needed to achieve electron localization was found to be on the order of few picoseconds [5]. Furthermore, as clear from Fig. 2(a) the localization is robust to considerable changes in the field strength which makes the present control scheme a good candidate for applications, such as the design of electro-optical devices and ultrafast switches.

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Self-interaction correction in multiple scattering theory: Application to cerium

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The self-interaction corrected local spin density (SIC-LSD) approximation [1, 2] has proved to be a useful scheme to describe static correlations in strongly correlated electron systems. In particular, it can determine whether an electron is delocalized or localized, i.e. whether its orbital is part of the valence states or not. This leads to a determination of the number of valence states and a nominal valence, as demonstrated by numerous calculations on rare earths, actinides, transition metal oxides, including the parent compounds of the high T_c materials and the CMR materials.

The full SIC-LSD scheme is unfortunately difficult to implement. This is due to the repeated transformations from reciprocal space (k-space) to real space to evaluate the self-interaction potential and the back transformations to k-space to solve the bandstructure problem. So far most applications of the full SIC formalism have been implemented in the LMTO-ASA band structure method [3]. A simpler scheme could lead to new functionality and applications such as full potential or an alloy description with the coherent potential approximation (CPA) [4, 5].

Motivated by these aims we have developed a new single-site SIC-LSD approach [6], referred to as local self-interaction correction (LSIC) formalism, which is an approximation to the full approach, but implemented in the multiple scattering theory. It is based on the experience with the full SIC-LSD formalism showing that to better than 98% the electron is localized on the site under consideration. A localized state has a very sharp resonance in its phase shift, associated with a large Wigner-delay time on a particular site. This allows us to make a single-site approximation to determine the SIC charge density and the SIC potential.

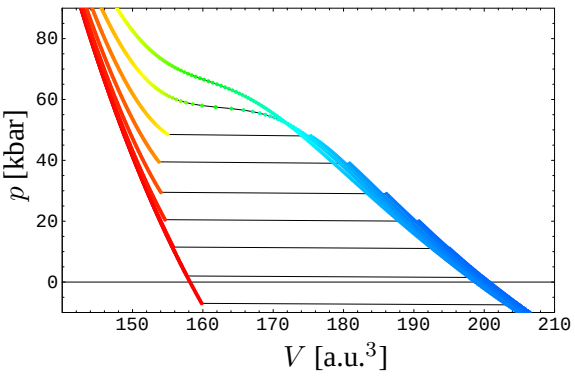

Fig. 1: Calculated isotherms for the temperatures T=0, 200, 400, 600, 800, 1000, 1200, 1400, and, 1600K. The color indicates the fraction of localized electrons: blue is all localized (γ-phase) and red is all delocalized (α-phase).

One of the advantages of the multiple scattering implementation of the SIC-LSD formalism is that it can be easily generalized to include the coherent potential approximation, [4, 5] extending the range of applications to random alloys. In addition, one can use it to treat static correlations beyond LSD by studying pseudoalloys whose constituents are composed e.g. of two different states of a given system: one delocalized, described by the LSD potential, and another localized, corresponding to the SIC-LSD potential. Combined with the DLM formalism for spin-fluctuations [7], this allows also for different orientations of the local moments of the constituents involved.

We have applied the SIC multiple scattering scheme in its CPA extension to the $\alpha - \gamma$

phase transition in Ce. Ce, being the first element containing an f electron, has an interesting phase diagram, showing, in particular, the iso-structural (fcc-fcc) $\alpha - \gamma$ phase transition, which is associated with a $15 - 17\%$ volume collapse and a quenching of the magnetic moment. The low-pressure γ-phase shows a local magnetic moment, and is associated with a trivalent configuration of Ce. At the temperatures in which the γ-phase is accessible, it is in a paramagnetic disordered local moment state. By increasing the pressure, the material first transforms into the α-phase, which is indicated to be an intermediate valence state with quenched magnetic moment. With increasing temperature, the α-phase transition shifts to higher pressures, ending in a critical point (600K, 20 kbar), above which there is a continuous crossover between the two phases. In order to determine the ground state configuration of Ce at the given volume, we have calculated the total energies for different volumes using the LDA for α-phase and the SIC formalism with correcting one f-electron, occupying in sequence all possible f-states, for the γ-phase. The calculated ground state properties of Ce are generally in good agreement with experiments, demonstrating the applicability of the SIC formalism. At finite temperatures we describe Ce as a pseudo-alloy of α and γ-Ce atoms. The calculated isotherms of Ce are displayed in figure 1. It can be seen that the average valence, close to the coexistence line, gradually changes with increasing temperature. Above the critical temperature, the valence changes continuously from trivalent to tetravalent with increasing pressure. In figure 2 we show the phase diagram, calculated from the free energies of the $\alpha - \gamma$-pseudo-alloy. It can clearly be seen in the figure how the transition becomes continuous above the critical temperature. The slope of the phase separation line is in very good agreement with the experiment. The critical temperature overestimates the experimental one by a factor of two, which is still reasonable considering that the critical temperature is very sensitive to various details of the calculation.

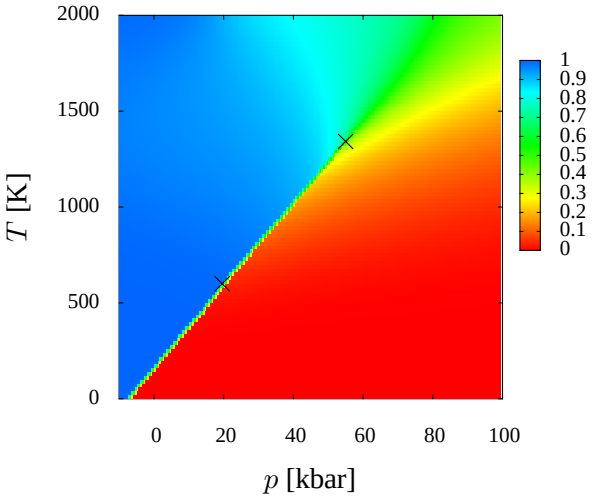

Fig. 2: Phase diagram obtained for the pseudoalloy, composed of α- and γ-Ce. The crosses indicate the calculated and experimental critical points.

The SIC multiple scattering formalism makes it possible to describe the ground state properties of a wide range of materials with localized electrons. In combination with the CPA it provides the basis for future developments, including a new dynamical mean field approach, which might be able to describe many-body effects such as the Kondo screening of local moments at low temperature.

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Topological Hall effect in magnetic nanostructures

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Possible physical mechanisms of anomalous Hall effect (AHE) in ferromagnets have attracted the attention of theorists for almost half a century [1] but the origin of AHE is still the subject of intensive dicussions [2]. Recently, some new important ideas have been brought into this field [3, 4, 6] relating the AHE to the topology of magnetization profile and to the Berry phase of electrons moving adiabatically in the inhomogeneous magnetic media. However, several key points of the topology-based analysis have been unclear like, for example, the role of spin-orbit (SO) interaction. In all previous theories including the chirality mechanism of AHE, the SO interaction was a necessary element.

We develop a new theory of the topological Hall effect (THE) [7] induced by the Berry phase of electrons or holes moving adiabatically in an inhomogeneous ferromagnet with nontrivial topology of the magnetization profile. We show that the Hall effect is nonvanishing even in the absence of SO interaction, and this effect has a purely topological origin. To discriminate the proposed mechanism of THE from any other mechanisms we suggest the design of a nanostructure with fully controllable parameters. The structure we propose consists of a periodic lattice of magnetic nanocylinders [8] on top of two-dimensional electron (hole) gas (2DEG) in the diluted magnetic semiconductor (DMS), see Fig. 1. Our main idea is to create the magnetization profile in the DMS in accordance with the orientation of stray field penetrating into DMS from the magnetic cylinders.

The starting point of our theoretical consideration is the Hamiltonian of a 2DEG in a smoothly varying magnetization field $\mathbf{M}(\mathbf{r})$

$$H = -\frac{\hbar^2}{2m}\frac{\partial^2}{\partial \mathbf{r}^2} - g\,\boldsymbol{\sigma}\cdot\mathbf{M}(\mathbf{r}), \qquad (1)$$

where the first term corresponds to the Hamiltonian of free electron gas, and g is the coupling constant. We assume that the amplitude

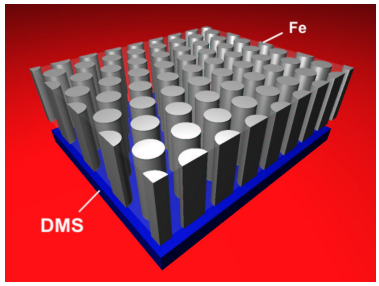

Fig. 1: The proposed structure consisting of a triangular lattice of magnetic nanocylinders on top of 2D diluted magnetic semiconductor.

of magnetization is constant, $\mathbf{M}(\mathbf{r}) = M\,\mathbf{n}(\mathbf{r})$, and that the 3D unit vector $\mathbf{n}(\mathbf{r})$ is a slowly varying function of coordinates. By apply-

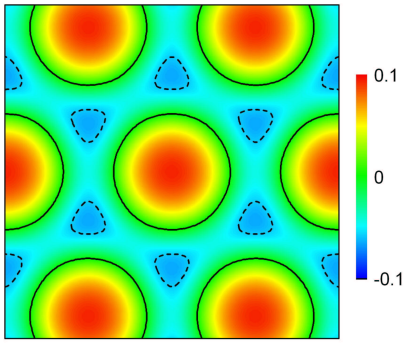

Fig. 2: Distribution of the z-component of dipolar field $B/4\pi M_s$ inside the semiconductor film for the triangular lattice of magnetic nanocylinders. The black solid circles are the lines of $B_z = 0$. Dashed lines correspond to the lines with $B_z = 0$ under an external field $B_{\rm ext}/4\pi M_s = +0.058$.

ing the local transformation of the Hamiltonian to the frame with quantization axis along the magnetization vector in each point of the space, in the adiabatic regime corresponding to a small rotation angle at the wavelength of electrons, we find that the system described by Hamiltonian (1) is equivalent to the 2DEG in homogenous magnetization, with an additional *topological* gauge field acting with the opposite sign on the spin up and down electrons. The topological field can be calculated using the magnetization profile $\mathbf{M}(\mathbf{r})$. This field acts on the carriers like a magnetic field via the

corresponding Lorenz force. The crucial finding of the theoretical analysis is that the average topological field is determined only by the topology of magnetization profile $\mathbf{M}(\mathbf{r})$.

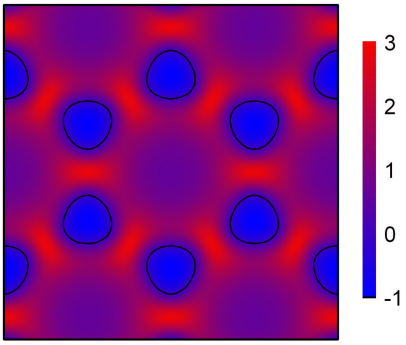

Fig. 3: Topological field $B_t(\mathbf{r})$ (in units of ϕ_0 per unit cell area) for the triangular lattice of magnetic nanocylinders. Black lines correspond to $B_t = 0$.

We calculated the magnetic field profile for the structure presented un Fig. 1. This field penetrates into DMS with a large density of magnetic impurities polarizing them in correspondence to the local dipolar field. The distribution of the field inside the semiconductor is shown in Fig. 2. The topological field B_t is related to the magnetization [7]

$$B_t = \frac{\phi_0}{4\pi}\,\epsilon_{\mu\nu\lambda}\,n_\mu\,(\partial_x n_\nu)\,(\partial_y n_\lambda), \qquad (2)$$

where ϕ_0 is the magnetic flux quantum, and $\epsilon_{\mu\nu\lambda}$ is the unit antisymmetric tensor. Using the results of our calculations of the dipolar field profile and Eq. (2), we found the topological field, which is presented in Fig. 3. For the structure with a triangular lattice of magnetic cylinders, the flux of topological field per unit cell equals to ϕ_0. Thus, the average field B_t does not vanish producing the Hall current.

If we apply an external magnetic field perpendicular to the DMS plane, it affects the topology of magnetization, and a new effect can be observed: the Hall conductivity vs. magnetic field reveals a nontrivial dependence at small magnetic fields (Fig. 4). In this figure, the sharp jumps correspond to the adiabatic limit but the real curve is smooth depending on the deviation from the adiabaticity. Using the typical parameters of the nanocylinder lattice and CdMnSe semiconductor, we analyzed the possibility to observe the THE for the DMS with hole density $n_p = 10^{11}$ cm^{-2} at $T = 4.2$ K. We found that the adiabaticity condition is fulfilled (the corresponding parameter is $\lambda = \varepsilon/(2gMk_F\xi) \simeq 0.04$, where ξ is of the order of the magnetic lattice constant).

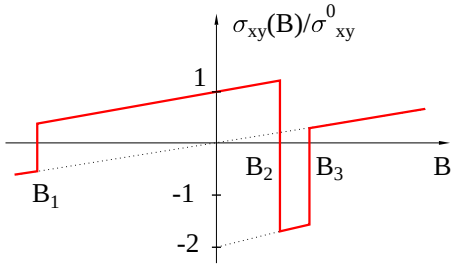

Fig. 4: Hall conductivity vs. external magnetic field (schematically). The slope corresponds to the contribution of the normal Hall effect; σ^0_{xy} is the Hall conductivity corresponding to a topological flux per unit cell equal to ϕ_0.

In conclusion, we demonstrated that the topological Hall effect can be observed in suitably chosen nanostructures, and that its striking behavior under an external magnetic field provides an unambiguous experimental signature of the THE.

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Preparation of ultrathin magnetic film systems by MBE and PLD

Bartsch, M.
Microprocesses of plastic deformation

Berakdar, J.
Electron correlation in solids

Berger, A.
Reaction kinetics in preceramic polymers

Bisio, F. 2 months
Optical surface second harmonic generation, time-resolved two-photon photoemission

Bouhassoune, M.
Ab-initio investigations of ferromagnetic/oxidic interfaces

Breitenstein, O.
Thermographic studies of semiconductors, investigation of defects in solar cells, electrical measurements on nanowires

Buonassisi, A. May – August
Investigation of defects in solar cells

Chelaru, L. I. until December
Photoemission microscopy, low-energy electron microscopy

Choi, J. until February
Self-organized structures in porous alumina

Christiansen, S. H.
Nanoscopic semiconductor layers and structures

Chu, M.-W. until January
Microstructure and properties of ferroelectric films, multilayers and mesoscopic structures

Cirlin, G. March – June
MBE growth of Si/Ge and InAs/GaAs structures

Daub, M. since January
Magnetic nanostructures for applications in biotechnology

Dietzsch, C.
High voltage electron microscopy, image analysis, dislocation structures

Dugaev, V.
Weak localization in ferromagnets

Edelmann, F. since September
Characterization of amorphous Si-layers in photovoltaics

Elhajal, M. since March
First-principles study of magnetic excitations in complex systems. Methods development and applications

Ernst, A.
First-principles study of surfaces and interfaces

Etzkorn, M. 10 months
Surface spin wave excitation by electrons

Fan, H.
Nanowire growth

Fecioru, A.
DLTS characterization of bonded semiconductor interfaces

Fleischer, F.
Pulsed laser deposition of nanowires

Franta, J. until June
Time-resolved two-photon photoemission, optical second harmonic generation

Fukumoto, K.
Photoemission microscopy and spectroscopy

Gai, Z. since May
Scanning electron microscopy with polarization analysis

Galanakis, I. July – August
Theory of magnetism

Gao, C. L.
Spin-polarized STM

Geppert, T. M.
Photonic crystal gas sensors based on macroporous silicon, IOSOI photonic crystals

Gerth, G.
Real structure of semiconductors, silicon MBE

Göring, P.
Nanoporous structures for biotechnology

Graff, A. until September
Solid state reactions on the nanometer scale

Ha, J. K. until January
Study of magnetic configurations and magnetization processes in submicron ferromagnetic elements using finite-element micromagnetic technique

Hähnel, A.
Nanokinetics of Si-C-O systems

Heitmann, J. until December
Si nanostructures, optical characterization

Hejjo Al-Rifai, M. until July
Lock-in thermography on solar cells

Henk, J.
Ab initio theory of scanning tunneling microscopy

Hertel, R.
Micromagnetism, finite-element modeling of magnetic structures and magnetization processes

Hesse, D.
Ferroelectric thin films and nanostructures, solid state reactions on the nanometer scale

Hillebrand, R.
Computations concerning photonic crystals

Höche, H.
UHV cantilever magnetometry

Hofmeister, H.
Nanoparticles

Hohlbein, J. 11 months
Protein folding in artificial cavities

Huang, R. since September
Elastic deformation at surfaces

Hünlich, K. since January
Correlation spectroscopy

Jamois, C. until August
Silicon-based photonic crystal slabs

Janke-Gilman, N. until August
Spin-polarized scanning tunneling microscopy

Ji, R. since March
Lithographically-controlled growth of porous alumina

Kabachnik, N. until February
Dynamics and correlation on the atomic scale

Karas, P. until May
First-principles study of surfaces and interfaces

Kästner, G.
Theoretical investigation of internal strains in nanowires

Kerherve, G. until January
Coincidence spectroscopy at surfaces

Kidun, O.
Electronic correlations in nanostructures

Kipke, A.
Magnetic resonance studies at silver-doped soda-lime glasses of varying iron content

Knez, M.
Synthesis, functionalization and characterization of nanowires

Kolb, F. M.
Nanowire growth

Kuch, W.
Magnetic dichroism, photoemission microscopy, photoelectron spectroscopy

Kudrnovský, J. April – June
First-principles studies of magnetism

Kuhlmann, V.
Development of empirical potentials for molecular dynamics simulations of bonded interfaces

Kumar, A. P. S. until May
Spin-polarized electron energy loss spectroscopy and hot-electron magnetotransport in spin-valve transistors

Kurochkina, E. May – July
Magnetic nanoclusters on surfaces

Laref, A. until May
Electronic properties of surfaces and interfaces of d and f transition metals

Ledig, L. until April
Plastic deformation of quasicrystals

Lee, S. K.
Structure-property relations of complex oxide thin films

Lee, W.
Electrodeposition of semiconductor nanowires

Lin, M.-T. July – August
Stress measurements of antiferromagnetic layers

Lisca, M. 6 months
Characterization of high-k dielectric and ferroelectric films

Liu, C.
Preparation of magnetic nanostructures by organic patterning

Liu, Z. January – September
Photoemission microscopy, antiferromagnetic thin films

Long, L. until December
PEW and SOI wafer bonding material based photosensors

Lotnyk, A. since April
Solid state reactions in electroceramic systems

Lu, T.
Excitation and transport of carriers in size-controlled silicon nanocrystals

Lu, Y. until March
Nuclear resonant scattering on Fe-layers on Cu(001) grown by pulsed laser deposition

Luo, F. since October
Growth of magnetic nanostructures by pulsed laser deposition

Luo, Y.
Tunable photonic crystals, complex one-dimensional nanostructures

Matos Abiague, A.
Theory of nano-scale optoelectronic devices

Matthias, S.
Ratchet systems and 3-D photonic crystals based on macroporous silicon

Mereu, B.
Integration of very high-k dielectrics with silicon technology

Messerschmidt, U.
Microprocesses of plastic deformation

Metalidis, G.
Electronic transport in mesoscopic systems

Meyerheim, H. L.
Growth, structure and magnetism of metal-metal and metal-oxide interfaces

Milenin, A.
Development of plasma processes for the patterning of nanostructures

Müller, F.
Porous materials

Navas, D. 5 months
Electrodeposition of ferromagnetic nanowires

Negulyaev, N. since September
Atomic nanostructures stabilized by surface-state electrons: Kinetic Monte Carlo simulations based on input from density functional theory

Niebergall, L. since January
Quantum electron confinement in metal nanostructures

Nielsch, K.
Multifunctional nanotubes and nanowires

Nyvlt, M.
Optical surface second harmonic generation, time-resolved two-photon photoemission

Pan, W. until August
Stress measurements during surface alloy formation

Pervak, V. March – May
Femtosecond laser spectroscopy

Pintilie, L.
Characterization of ferroelectric thin films

Pippel, E.
High resolution and analytical electron microscopy, ELNES, EFTEM

Polok, M. since April
Correlation between magnetism, structure and topology of thin films

Ponomareva, I. August – September
Nonequilibrium transport in nano-size materials

Popescu, R.
Surface X-ray diffraction analysis on ultra-thin films

Przybylski, M.
Correlation between magnetism, structure and topology of thin films; Fe-based spin aligners epitaxially grown on semiconducting substrates

Radu, I.
Layer splitting by ion implantation and direct wafer bonding

Rakotoniaina, P.
Investigation of defects in solar cells

Rehn, U.
Bioconjugated metal nanoparticles

Reiche, M.
Wafer bonding

Reischer, F.
Interlayer design in metal matrix composites

Richter, S.
Optically active elements in photonic crystals

Rico Hernandez, M. until March
Layer transfer and wafer bonding

Sander, D.
Mechanical stress, magnetostriction, and magnetic anisotropy of ferromagnetic layers

Sandratskii, L.
First-principles investigations of magnetic materials

Sasioglu, E.
First-principles calculation of exchange interactions and Curie temperature of complex magnetic systems

Scheerschmidt, K.
Interfaces: Molecular dynamics simulation of structures and inverse object retrieval

Schicke, K.-D.
Metal particles in glass

Schlickum, U. 11 months
Spin-polarized STM

Schlitt, St. since February
Polymere Nanoröhrchen

Schmidt, V.
Semiconductor nanowires

Schneider, R.
Nanoparticles in glasses

Scholz, R.
TEM of interfaces, multilayers and nanomaterials

Schubert, L.
Growth and properties of thin InAs layers on Si

Senz, S.
UHV wafer bonding, spin-valve transistor, semiconductor nanowires

Shen, J. since February
CVD growth of Si nanocrystals

Shi, Y. since February
Correlation between magnetism, structure and topology of thin films

Singh, R. since January
Fabrication and characterization of nanomechanical cantilevers through wafer bonding and SOI approaches

Slowik, B. until April
Spin-polarized scanning tunneling microscopy

Sokolov, L. since September
The influence of submonolayers of carbon on the Stranski-Krastanov growth of SiGe layers

Sommerlatte, J. since September
Fabrication of thermoelectric nanowires

Soyka, O. since March
Surface X-ray diffraction

Steinhart, M.
Complex one-dimensional nanostructures, structure formation in one-dimensional confinement

Stepanyuk, V. S.
Magnetic nanostructures

Sun, Y. since April
Block copolymer nanowires

Sun, Yan since September
Introduction to the electron beam lithography

Sundermann, A. since January
Molecular design of precursors, band structure calculations for ELNES interpretations, modelling of reaction kinetics

Taillefumier, M.
Berry phase and Hall effect in nanostructures

Talalaev, V.
Optical properties of Ge and Si nanoclusters

Tang, W.
Surface spin wave excitation by electron energy loss spectroscopy

Tian, Z. since August
Surface magnon spectroscopy

Tieg, C.
Spin-resolved photoelectron spectroscopy, photoelectron diffraction

Tonkikh, A. 4 months
MBE-growth of SiGe islands

Tusche, C. since April
Growth, structure and magnetism of metal-oxide interfaces

Utsumi, Y. until June
Theoretical study of the microscopic mechanism of spin transfer torque

Vollmer, R.
Spin-polarized electron energy loss spectroscopy

Wang, S.
Spin-resolved photoelectron spectroscopy, photoelectron diffraction

Wang, W. until March
Epitaxial growth of half-metallic thin films by pulsed laser deposition

Wasniowska, M.
Spin-polarized scanning tunneling spectroscopy

Werner, M.
TEM of multicrystalline materials for solar cells

Werner, P.
High resolution electron microscopy, nanostructures in semiconductor materials, growth of Si-heterostructures

Winkelmann, A. since March
Magnetooptic Kerr effect studies of thin films

Winkler, C.
Low energy electron coincidence spectroscopy

Woltersdorf, J.
Optimization of interfaces and nanokinetics of metallic and ceramic systems

Wu, Ch. July – September
Spin-polarized STM on antiferromagnets

Wulfhekel, W.
Spin-polarized STM, spin-polarized transport and magnetoelectronics

Yamasaki, A. since September
Spin-polarized STM on iron

Yan, L. until August
Correlation between magnetism, structure and topology of thin films

Yu, G. until March
Chemical and magnetic characterization of FM/NiO interfaces

Yu, J. March – September
Electroceramic nanopowders

Zacharias, M.
Si-based photonic structures and nanocrystals

Zakharov, N.
(HR)TEM investigations of nanostructures

Zhang, L. since June
Spin polarized scanning electron microscopy

Zhang, Y. since August
Surface magnon spectroscopy

Zhao, H. July – September
Study of surface magnons on Co films

Zhao, L.
One-dimensional compound semiconductor nanostructures

Zhu, X.
TEM of ferroelectric thin films and multilayers

Zschech, D.
Lithographic polymer templates for ordered porous alumina

Zukrowski, J. since April
Growth of ultrathin magnetic films

Scientists from abroad

Country	Number
PR China	21
Russia	12
Romania	9
Poland	5
Czech Republic	4
France	4
Ukraine	4
Japan	3
Korea	3
Taiwan	3
India	2
Spain	2
USA	2
Algeria	1
Belgium	1
Croatia	1
Cuba	1
Italy	1
Madagascar	1
Morocco	1
New Zealand	1
Syria	1
Turkey	1

Third-party funds

BMBF Projects

Bruno, P.
Spinelektronik in Halbleitern, Teilvorhaben: Theoretische Untersuchung verdünnter magnetischer Halbleiter
term of contract: 01.03.2002 – 28.02.2005
funding: EUR 267 000

Bruno, P.
Elektronenkorrelation und Dissipationsprozesse in Halbleiterquantenstrukturen, Theorie des spinabhängigen Elektronentransports in auf III-V-Halbleitern basierenden Spin-Elektronik-Systemen
term of contract: 01.01.2000 – 30.06.2004
funding: DEM 377 344

Kirschner, J.
Detektor für die Abbildung magnetischer Nanostrukturen im UHV SEM
term of contract: 01.04.1999 – 30.04.2004
funding: EUR 494 810

Kirschner, J.
Struktur, Magnetismus und spinpolarisierter Transport in FM/SC Heterostrukturen
term of contract: 01.10.2002 – 30.09.2005
funding: EUR 17 492

Kuch, W.
Unterauftragnehmer im Projekt: Zeit- und lagenaufgelöste Magnetisierungsdynamik in magnetischen Mehrschichtsystemen
term of contract: 01.07.2004 – 30.06.2007
funding: 1 scientific coworker

Kuch, W.
Elementspezifische Abbildung magnetischer Domänen in gekoppelten magnetischen Dünnschichtsystemen
term of contract: 01.04.2001 – 31.03.2004
funding: DEM 171 400

Nielsch, K.
Multifunktionale Nanostäbe und Nanoröhren
term of contract: 01.10.2003 – 30.09.2008
funding: EUR 1 741 171

Reiche, M.
Mikrosystembaustein zur intelligenten Ansteuerung von Motoren - MIBAUM, Teilvorhaben: Charakterisierung von Substraten für intelligente Motorsteuerungen
term of contract: 01.01.2001 – 31.07.2004
funding: EUR 180 664

Wehrspohn, R. B.
Nanobiotechnologie-Verbundprojekt der Westfälischen Wilhems-Universität Münster, High-throughput Analyseverfahren zum Screening des Metabolit- und Pharmakatransports über die Blut-Liquor-Schranke auf Basis geordneter Nanoporen
term of contract: 01.11.2001 – 31.10.2005
funding: EUR 289 251

Wehrspohn, R. B.
Planare photonische Kristalle in Materialsystemen mit hohem Brechzahlkontrast für die Telekommunikation, Herstellung von photonischen Kristall-Komponenten für die Dispersionskompensation auf SOI Basis
term of contract: 01.02.2002 – 31.01.2005
funding: EUR 545 548

DFG Projects

Ernst, A.
Forschergruppe Oxidische Grenzflächen, Teilprojekt B 3: Oxid-Ferromagnet-Grenzflächen für die Magnetoelektronik
term of contract: 01.07.2004 – 30.06.2005
funding: 1 scientific coworker BAT-O IIa

Hesse, D.
Forschergruppe Oxidische Grenzflächen, Teilprojekt B 1: Einfluß innerer Grenzflächen auf die Umschaltdynamik ferroelektrischer Oxidschichten
term of contract: 01.01.2003 – 31.12.2003
funding: 1 scientific coworker BAT-O IIa/2

Hesse, D.
Forschergruppe Oxidische Grenzflächen, Teilprojekt B 1: Einfluß innerer Grenzflächen auf die Umschaltdynamik ferroelektrischer Multilagen
term of contract: 01.01.2004 – 31.12.2005
funding: 1 scientific coworker BAT-O IIa 3/4, 1 scientific coworker BAT-O IIa/2

Kuch, W.
Forschergruppe Oxidische Grenzflächen, Teilprojekt A 10: Röntgenabsorptionsspektroskopie von magnetischen Metall-/antiferromagnetischen Oxid-Grenzflächen
term of contract: 01.01.2004 – 31.12.2005
funding: 1 scientific coworker BAT-O IIa 3/4

Messerschmidt, U.
Quasikristalle: Struktur und physikalische Eigenschaften; Plastische Verformung von Ein-Quasikristallen
term of contract: 01.09.2001 – 30.04.2004
funding: 1 scientific coworker BAT-O IIa for 22 months

Meyerheim, H. L.
Forschergruppe Oxidische Grenzflächen, Teilprojekt A 6: Magnetische, strukturelle und chemische Eigenschaften oxidischer Grenzflächen und Filme
term of contract: 16.04.2004 – 15.04.2006
funding: 1 scientific coworker BAT-O IIa 3/4

Sandratskii, L. M.
Schwerpunktprogramm: Moderne und universelle first-principles Methoden für Mehrelektronensysteme in Chemie und Physik, Thema: First-principles method for the calculation of magnons in real materials
term of contract: 01.03.2004 – 31.08.2005
funding: 1 scientific coworker BAT-O IIa

Steinhart, M.
Schwerpunktprogramm: Nanodrähte und Nanoröhren - Von kontrollierter Synthese zur Funktion, Thema: Coumpound semiconductor 1D nanostructures by template wetting
term of contract: 01.06.2004 – 31.05.2006
funding: 1 scientific coworker BAT-O IIa/2 for two years

Stepanyuk, V. S.
Magnetism, structure and interactions on the atomic scale: Supported metal clusters on metal surfaces
term of contract: 01.01.2004 – 31.12.2005
funding: 1 scientific coworker BAT-O IIa 3/4, EUR 1 500 other costs

Stepanyuk, V. S.
Schwerpunktprogramm: Nanodrähte und Nanoröhren - Von kontrollierter Synthese zur Funktion, Thema: Structure, electronic and magnetic properties of metal nanowires
term of contract: 01.06.2004 – 31.05.2006
funding: 1 scientific coworker BAT-O IIa/2

Woltersdorf, J.
Optimierung von Metallmatrix-Verbundstoffen durch Grenzschichtdesign
term of contract: 01.08.2004 – 31.07.2006
funding: 1 scientific coworker BAT-O IIa

Woltersdorf, J.
Optimierung von Metallmatrix-Verbundwerkstoffen durch Grenzschichtdesign
term of contract: 01.04.2002 – 31.03.2005
funding: 1 scientific coworker BAT-O IIa, DEM 6 000 consumables

Woltersdorf, J.
Neue oxidationsstabile und kriechfeste Si(Al,B)CO-Gradientenfasern aus modifizierten Polysiloxanen
term of contract: 16.07.2004 – 15.01.2006
funding: 1 scientific coworker BAT-O IIa

Woltersdorf, J.
Schwerpunktprogramm: Neue Precursorkeramik aus kondensierten molekularen Vorstufen, Thema: Einfluß partikulärer Füllstoffe auf das Pyrolyseverhalten präkeramischer Polymere
term of contract: 01.08.2004 – 31.07.2005
funding: 1 scientific coworker BAT-O IIa

Woltersdorf, J.
Schwerpunktprogramm: Precursorkeramik, Thema: Molekulares Design präkeramischer Polymere durch quantenchemische Verfahren am Beispiel der Synthese neuartiger Metallcarbodiimid-Polymere und -Gele
term of contract: 01.01.2002 – 15.07.2004
funding: 1 scientific coworker BAT-O Ib

Woltersdorf, J.
Schwerpunktprogramm: Neue Precursorkeramik aus kondensierten molekularen Vorstufen, Thema: Katalytische Steuerung von Vernetzung und Ausbeute
term of contract: 01.05.2003 – 30.04.2005
funding: 1 scientific coworker BAT-O IIa

Zacharias, M.
Größenkontrollierte Si-Nanokristalle für photonische und elektronische Anlagen
term of contract: 01.07.2004 – 30.06.2006
funding: 1 scientific coworker BAT-O IIa 3/4

Zacharias, M.
Photonische Strukturen auf Siliziumbasis - Herstellung, Charakterisierung und physikalische Eigenschaften von Siliziumnanokristallen definierter Größe
term of contract: 01.01.2002 – 31.12.2003
funding: 1 scientific coworker BAT-O IIa 3/4

Zacharias, M.
Schwerpunktprogramm: Photonische Kristalle, Thema: Optically active materials in photonic crystals
term of contract: 01.04.2002 – 30.04.2005
funding: 1 scientific coworker BAT-O IIa, DEM 19 500 materials, DEM 8 000 travelling costs, DEM 16 500 other costs

Zacharias, M.
Schwerpunktprogramm: Nanodrähte und Nanoröhren - Von kontrollierter Synthese zur Funktion, Thema: Controlled synthesis of Si nanowires and the realization of heterostructures by vapor-liquid-solid growth in combination with laser ablation
term of contract: 01.06.2004 – 31.05.2006
funding: 1 scientific coworker BAT-O IIa 3/4-Stelle

Zacharias, M.
Schwerpunktprogramm: Nanodrähte und Nanoröhren - Von kontrollierter Synthese zur Funktion, - Koordinator -
term of contract: 01.06.2004 – 31.05.2006
funding: EUR 20 000 for student assistant, EUR 5 000 for consumables

Special Investigation Projects (DFG Sonderforschungsbereiche)

Hesse, D.
Struktur und Dynamik nanoskopischer Inhomogenitäten in kondensierter Materie (SFB 418 MLU Halle-Wittenberg - MPI-MSP), Teilprojekt: Reaktionsfronten bei der $BaTiO_3$-Genese durch Festkörperreaktionen mit mikroskopischer Reaktionsgeometrie
term of contract: 01.01.2003 – 31.12.2005
funding: EUR 176 800

Hofmeister, H.
Struktur und Dynamik nanoskopischer Inhomogenitäten in kondensierter Materie (SFB 418 MLU Halle-Wittenberg - MPI-MSP), Teilprojekt: Struktur-Eigenschafts-Korrelationen nanoskaliger Metallteilchen in Gläsern
term of contract: 01.01.2003 – 31.12.2005
funding: EUR 36 000

Woltersdorf, J.
Struktur und Dynamik nanoskopischer Inhomogenitäten in kondensierter Materie (SFB 418 MLU Halle-Wittenberg - MPI-MSP), Teilprojekt: Reaktionskinetik im Nanometer-Bereich und ihr Einfluss auf die Eigenschaften von Verbundsystemen
term of contract: 01.01.2003 – 31.12.2005
funding: EUR 168 600

Sachsen-Anhalt Ministry of Education and Culture

Gösele, U.
Halbleiterbasismaterialien der flexiblen Elektronik: Entwicklung neuer Herstellungs- und Charakterisierungsverfahren für dünnes und ultradünnes Silizium (ULTRA-SI)
term of contract: 01.01.2004 – 31.12.2004
funding: EUR 104 304

Matthias, S.
Herstellung und Anwendung von 3-dimensionalem makroporösem Silizium
term of contract: 01.01.2004 – 31.12.2004
funding: EUR 50 000

Wulfhekel, W.
Rastertunnelmikroskop für magnetische Strukturen
term of contract: 01.01.2002 – 31.12.2003
funding: EUR 51 129

VW-Foundation

Alexe, M.
Nano-sized ferroelectric hybrids
term of contract: 04.07.2001 – 31.12.2004
funding: EUR 131 000

Reiche, M.
Self-organized pattern formation of biomolecules at silicon interfaces (SOBSI)
term of contract: 01.05.2004 – 30.04.2007
funding: EUR 145 800

Zacharias, M.
Er-Dotierung von Si-Nanostrukturen und deren optische und strukturelle Untersuchung
term of contract: 13.09.2000 – 31.01.2004
funding: DEM 349 640

EU Projects

Alexe, M., U. Gösele
Single-Crystalline Thin Films by Direct Wafer Bonding and Hydrogen Induced Exfoliation
term of contract: 18.08.2000 – 31.12.2005
funding: EUR 295 200 Marie Curie Fellowship

Alexe, M.
Waferbonding and Active Passive Integration Techology and Implantation
term of contract: 01.06.2004 – 31.05.2007
funding: EUR 145 000

Breitenstein, O.
Solar Cell Performance Optimisation Relating Process Tracking by Imaging Techniques with Modelling (PORTRAIT)
term of contract: 01.01.2002 – 31.12.2004
funding: EUR 172 368

Bruno, P.
Computational Magnetoelectronics Research Training Network
term of contract: 01.10.2000 – 30.09.2004
funding: EUR 199 800

Gösele, U., M. Alexe
INVEST, Integration of Very High-k Dielectrics with Silicon CMOS Technology
term of contract: 01.07.2001 – 31.10.2004
funding: EUR 144 624

Werner, P.
Coordination: Action on Defects Relevant to Engineering Advanced Silicon-Based Devices
term of contract: 01.01.2004 – 31.12.2006
funding: EUR 36 000

Zacharias, M.
INTAS, Erbium-Doped Heterogeneous Structures on Silicon: Basis for Optoelectronic Applications
term of contract: 01.03.2004 – 28.02.2006
funding: EUR 7 750

BMWI Projects

Breitenstein, O.
Alternatives Silicium für Solarzellen (ASIS), Teilprojekt 5: Untersuchung von Shunts und Mikrorissen mittels Lock-in-Thermographie und Transmissions-Elektronenmikroskopie
term of contract: 01.10.2002 – 30.09.2005
funding: EUR 386 108

German Academic Exchange Service

Kuch, W.
Wachstum und magnetische Eigenschaften von epitaktischen Legierungsschichten und Schichtstapeln
term of contract: 01.01.2004 – 31.12.2005
funding: EUR 5 901

Wulfhekel, W.
Geordnetes Wachstum magnetischer Nanoteilchen auf kontrolliert vorstrukturierten Siliziumkarbidsubstraten
term of contract: 01.01.2004 – 31.12.2005
funding: EUR 10 598

Miscellaneous / Industrial Funding

Christiansen, S. H., M. Reiche, U. Gösele
Entwicklung und Charakterisierung von SSOI-Wafern (Siltronic AG, München)
term of contract: 01.05.2004 – 30.04.2007
funding: EUR 537 200

Gösele, U., S. H. Christiansen
Entwicklung spezifischer Waferbondtechniken und Bestimmung geeigneter Wasserstoffimplantationsbedingungen zur Übertragung dünner einkristalliner GaN-Schichten auf Saphirsubstrate (OSRAM Opto Semiconductors GmbH, Regensburg)
term of contract: 01.01.2004 – 04.09.2004
funding: EUR 60 000

Pippel, E., J. Woltersdorf
Characterization of ion-irradiated sol-gel materials (CNRS, France)
term of contract: 01.10.2003 – 31.10.2004
funding: EUR 5000

Reiche, M.
Entwicklung von Waferbondtechniken für photonische Mikrosysteme (Fraunhofer-IPMS, Dresden)
term of contract: 01.04.2004 – 31.12.2004
funding: EUR 17 500

Reiche, M.
Entwicklung spezifischer Waferbondtechniken (SÜSS MicroTec AG, Garching)
term of contract: 01.10.2003 – 30.09.2005
funding: EUR 110 000

Werner, P.
Strukturelle Charakterisierung von Quantenpunktschichten (Institut für Festkörperforschung, TU Berlin)
term of contract: 01.04.2000 – 31.12.2003
funding: DEM 177 570

Doctoral, habilitation and diploma theses

Dissertations

Matos Abiague, A.
Dynamics of quantum systems driven by half-cycle electromagnetic pulses
12.10.2004
MPI

Richter, S.
Periodische Punktdefektstrukturen und Quantenpunktemitter in zweidimensionalen photonischen Kristallen
06.07.2004
DFG

Jamois, C.
Silicon-based planar photonic crystals for application to dispersion compensation
16.06.2004
BMBF

Choi, J.
Fabrication of monodomain porous alumina using nanoimprint lithography and its applications
05.02.2004
MPI

Chelaru, L. I.
Microscopic studies of interlayer magnetic coupling across nonmagnetic and antiferromagnetic spacer layers
16.12.2003
MPI

Liu, G.
Exchange interaction of Fe films on NiO(001) single crystals
09.12.2003
MPI

Ledig, L.
Temperaturabhängigkeit der Versetzungsstrukturen in plastisch verformten Al-Pd-Mn-Ein-Quasikristallen
03.12.2003
MPI

Awards

Lee, S. K., Lee, H. N., Zakharov, N. D., Zhu, X. H., Hesse, D.
- Best Poster Award, 16th International Symposium on Integrated Ferroelectrics (ISIF 2004)

Sander, D.
- Gaede-Preis der Deutschen Vakuumgesellschaft

Steinhart, M.
- Scientific Award of the Industry Association of Kassel for excellent scientific work at the Philipps University Marburg

Szafraniak, I., Chu, M.-W., Alexe, M., Hesse, D.
- Best Poster Award, 8th International Symposium on Ferroic Domains (ISFD-8)

Appointments as professor

Bao, D.
- Full professorship at the Sun Yat-sen University, Guangzhou, PR China, accepted

Kuch, W.
- C3-Professur für Experimentelle Festkörperphysik, Fachbereich Physik, Freie Universität Berlin

Ma, W.
- Professorship at Shantou University, Guangdong, PR China, accepted

Sander, D.
- Vertretungsprofessur WS 2004, Universität Erlangen-Nürnberg

Wulfhekel, W.
- C3-Professur für Experimentalphysik, Universität des Saarlandes

Invited professorship

Dugaev, V. K.
- University of Grenoble

Activities in scientific boards

Academies, scientific societies, committees etc.

Berakdar, J.
- Member of the Scientific Committee of the International Conference on Electron and Photon Impact Ionization and Related Topics

Bruno, P.
- Jurymitglied: "Forschungspreis" und "Preis für Angewandte Forschung" des Landes Sachsen-Anhalt
- Member of the Comité Scientifique Laboratoire de l'État Condensé, CNRS (Le Mans, France)
- Member of the Expert Panel, Mission Scientifique Technique et Pédagogique, Ministere de la Recherche, France

Gösele, U.
- Board of Directors, Materials Research Society, USA
- Fellow of the American Institute of Physics, USA
- Fellow of the Institute of Physics, United Kingdom
- Innovationsrat des Landes Sachsen-Anhalt
- Member, Advisory Board, Research Center Rossendorf
- Member in the Reviewing Boards of the Helmholtz Society programs "Information Technology with Nanoelectronic Systems" at the Research Center Jülich and "Nanotechnology" at the Research Center Karlsruhe
- Mitglied der Deutschen Akademie der Naturforscher Leopoldina
- Mitglied der Kaiserlich-Königlichen Böhmischen Physikalischen Gesellschaft

Hesse, D.
- Member of the Chemistry, Physics and Technology Section of the Scientific Council of the Max Planck Society

Hofmeister, H.
- Advisory Board of the International Centre for Advanced Materials Science and Electron Microscopy

- Member of the Organizing Committee of NATO Advanced Study Institute "Nanostructured and Advanced Materials for Optoelectronic, Photovoltaic and Sensor Applications"

Kirschner, J.

- Member of the Research Center "International Center for Quantum Structures" (Chinese Academy of Sciences), Beijing, PR China
- Mitglied der Deutschen Akademie der Naturforscher Leopoldina
- Mitglied im Wissenschaftlichen Beirat des Hahn-Meitner-Instituts, Berlin
- Mitglied im Wissenschaftlichen Beirat des Instituts für Schichten und Grenzflächen (ISG) des Forschungszentrum Jülich

Kuch, W.

- Vice-Chairman of the BESSY User Committee

Meyerheim, H. L.

- Mitglied im Proposal Review Committee der ESRF für den Bereich "Surfaces/Interfaces"

Pippel, E.

- Mitglied des Arbeitskreises "Energiefilterung und Elektronen-Energieverlust-Spektroskopie" der DGE
- Mitglied des Arbeitskreises "Polymerkeramik" des DGM / DKG - Gemeinschaftsausschusses "Hochleistungskeramik"

Sander, D.

- Member of the Review Panel of Beamline ID-03 at ESRF, Grenoble

Scheerschmidt, K.

- Sprecher des Arbeitskreises "Hochauflösende Elektronenmikroskopie" der DGE

Woltersdorf, J.

- Leiter des Graduierten-Seminars Physik der Martin-Luther-Universität Halle-Wittenberg und des Max-Planck-Instituts für Mikrostrukturphysik
- Mitglied des Arbeitskreises "Polymerkeramik" des DGM / DKG - Gemeinschaftsausschusses "Hochleistungskeramik"
- Mitglied des Arbeitskreises "Verstärkung keramischer Werkstoffe" im DGM / DKG - Gemeinschaftsausschuss "Hochleistungskeramik"
- Mitglied des Beraterkreises des Vorstandes der Deutschen Gesellschaft für Materialkunde e.V.
- Mitglied des DGG-DKG-Arbeitskreises "Glasig-kristalline Multifunktionswerkstoffe"
- Mitglied des DGM-Arbeitskreises "Precursorsynthese"
- Mitglied des Preiskuratoriums der Deutschen Gesellschaft für Materialforschung
- Mitglied des Programmausschusses des DFG-Schwerpunktprogramms "Nanostrukturierte anorganische Materialien durch molekulares Design: Werkstoffe für neue Technologiefelder"
- Stellv. Sprecher des Sonderforschungsbereiches "Struktur und Dynamik nanoskopischer Inhomogenitäten in kondensierter Materie" der Martin-Luther-Universität Halle-Wittenberg und des Max-Planck-Instituts für Mikrostrukturphysik

Zacharias, M.

- Hauptkoordinatorin des DFG-Schwerpunktprogramms 1165 "Nanodrähte und Nanoröhren: Von kontrollierter Synthese zur Funktion"

Publishing committees of scientific journals

Gösele, U.

- Advisory Board "Advanced Functional Materials"
- Advisory Board "Zeitschrift für Metallkunde"
- Editorial Board "Applied Physics A"
- Editorial Board "Information Sciences"

Kirschner, J.

- Advisory Board "Journal of Magnetism and Magnetic Materials"
- Guest Editor of Special Issue on "Vacuum based science and technology", Applied Physics A
- Member of the Editorial Advisory Board for the "Encyclopedia of Nanoscience and Nanotechnology"

Sander, D.

- Applied Surface Science (Guest Editor)

Preparing committees of conferences

Alexe, M.

- Member of European Steering Committee on Ferroelectricity Meetings (ESCF)
- Permanent Member of the International Advisory Board of the European Meeting on Ferroelectrics

Berakdar, J.

- International Conference on Electron and Photon Impact Ionization and Related Topics
- Member of the International Advisory Committee of the "International Symposium on (e,2e), Double Photoionization and Related Topics" to be held in Buenos Aires, Argentina, July 28-30, 2005, as a satellite of the XXIX ICPEAC Conference taking place in Rosario, Argentina, July 20-26, 2005

Breitenstein, O.

- Member of the International Scientific Committee 'Beam Injection Assessment of Microstructures in Semiconductors' (BIAMS)

Bruno, P.

- Chair of the Gordon Conference on Magnetic Nanostructures, Big Sky, USA
- Co-Chair of the WE Heraeus Summer School on Spin-Electronics, Wittenberg, Germany
- Co-Director of the Workshop on Magnetic Nanostructures, Brasilia, Brazil
- International Conference on Nanospintronics Design and Realization (ICNDR)
- Member of the International Advisory Committee of the 4th Mittelwihr School on Magnetism and Synchrotron Radiation, Mittelwihr, France

- Member of the International Advisory Committee of the International Conference on Magnetism, Kyoto (Japan) 2006
- Member of the International Advisory Committee of the Moscow International Symposium on Magnetism (MISM-2005), Moscow, Russia
- Member of the International Advisory Committee of the Third International Conference and School on Spintronics and Quantum Information Technology (SPINTECH III), Awaiji Island, Japan
- Member of the Program Committee of the Ecole "Rayonnement Synchrotron et Magnétisme"

Gösele, U.

- Co-Organizer WE-Heraeus School "Science and Technology of Inorganic Nanowires and Nanotubes"
- Member of Scientific Advisory Board "10th European Symposium on Semiconductor Detectors"
- Member of Scientific Advisory Board "Workshop on Wafer Bonding for MEMS Technologies"
- Member of the Advisory Board of the Autumn School on Materials Science and Electron Microscopy 2004 "Emerging Microscopy for Advanced Materials Development - Imaging and Spectroscopy on Atomic Scale", International Centre of Advanced Materials Science and Electron Microscopy
- Member of the Scientific Advisory Board of Symposium I "Current Trends in Nanoscience - from Materials to Applications" of the European Materials Research Society Spring Meeting (E-MRS) 2004
- Member of the Scientific Advisory Board of the Electronic Materials Conference 2004
- Semiconductor Silicon 2006

Hesse, D.

- Co-Organizer of the Interdisciplinary Symposium on Non-Volatile Solid State Memories, 69. Jahrestagung der DPG "Physik seit Einstein"
- Co-Organizer of the Symposium on Functional Oxide Epitaxial Thin Films and Multilayers, 14th International Conference on Crystal Growth (ICCG-14)
- Co-Organizer of the Symposium on Solid State Chemistry, Dreiländertagung "Microscopy Conference 2005"
- Member of the International Advisory Comittee, International Conference on Solid-Solid Phase Transformations (PTM '05)
- Member of the Scientific Committee, EuroConference on Interfaces in Nanostructured Materials

Hofmeister, H.

- Member of the Advisory Board of the Autumn School on Materials Science and Electron Microscopy 2004 "Emerging Microscopy for Advanced Materials Development - Imaging and Spectroscopy on Atomic Scale", International Centre of Advanced Materials Science and Electron Microscopy
- Organizing Committee of the NATO Advanced Study Institute "Nanostructured and Advanced Materials for Optoelectronic, Photovoltaic and Sensor Applications"

Kirschner, J.

- Member of the International Advisory Committee of the 4th International Symposium on Metallic Multilayers (MML '04)
- Member of the International Advisory Committee of the Joint European Magnetic Symposia (JEMS 2004)
- Member of the Program Committee of the 9th Joint MMM/Intermag Conference

Sander, D.

- Member of the Preparing Committee of the 14th International Conference on Crystal Growth (ICCG-14)

Scientific events

Scientific meetings

PORTRAIT Project Meeting
Halle, June 24–25, 2004

Workshop "Physikalische Fehleranalyse an hochintegrierten mikroelektronischen Schaltkreisen: Stand und zukünftige Herausforderungen"
Halle, September 10, 2004 (with Fraunhofer-Institut für Werkstoffmechanik Halle and Martin-Luther-Universität Halle-Wittenberg)

WE-Heraeus-Ferienkurs für Physik "Physik der magnetischen Nanostrukturen"
Halle, September 13–24, 2004 (with Martin-Luther-Universität Halle-Wittenberg)

Joint European Laboratory (Laboratoire Européen Associé - LEA) Meeting
Halle, September 30 – October 01, 2004

Autumn School 2004: Emerging Microscopy for Advanced Materials Development - Imaging and Spectroscopy on the Atomic Scale
Berlin, October 03–07, 2004 (with International Centre of Advanced Materials Science and Electron Microscopy)

Workshop on Wafer Bonding for MEMS Technologies
Halle, October 10–12, 2004 (with Fraunhofer-Institut für Werkstoffmechanik Halle)

Joint colloquia with the Physics Department at the Martin Luther University Halle-Wittenberg

S. Großmann (04.12.2003)
Phillipps-Universität Marburg, Institut für Physik, Marburg
Strömungsturbulenz und wie sie entsteht

F. Jülicher (11.12.2003)
Max-Planck-Institut für Physik komplexer Systeme, Dresden
Physik des Hörens: Aktive Schallwahrnehmung im Innenohr

T. Thurn-Albrecht (18.12.2003)
Martin-Luther-Universität Halle-Wittenberg, Fachbereich Physik, Halle
Vom Zufallsknäuel zur selbstorganisierten Nanostruktur - ein Streifzug duch die Physik der Polymere

C. Hübner (05.02.2004)
Martin-Luther-Universität Halle-Wittenberg, Fachbereich Physik, Halle
Von Nano-Antennen und einzelnen Photonen

H. Löwen (15.04.2004)
Heinrich-Heine-Universität Düsseldorf, Institut für Theoretische Physik II, Düsseldorf
Was ist hart an "Weicher Materie"?

A. Bunde (29.04.2004)
Justus-Liebig-Universität Gießen, Institut für Theoretische Physik III, Gießen
Langzeitkorrelationen in der Natur

H. Lichte (06.05.2004)
Technische Universität Dresden, Institut für Strukturphysik, Dresden
Properties of electron waves and applications in solid state physics

G. Hasinger (03.06.2004)
Max-Planck-Institut für extraterrestrische Physik, Garching
Das Wachstum der Schwarzen Löcher im Universum

M. F. de Kieviet (10.06.2004)
Universität Heidelberg, Physikalisches Institut, Heidelberg
Lots of whirl about the vacuum? A quantitative experiment on the Casimir force

D. P. van der Werf (01.07.2004)
CERN / University of Wales, Swansea, United Kingdom
ATHENA - The Antihydrogen Experiment at CERN

J. Kantelhardt (15.07.2004)
Martin-Luther-Universität Halle-Wittenberg, Fachbereich Physik, Halle
Phasenübergänge in Elektronensystemen

F. Gebhard (21.10.2004)
Phillipps-Universität Marburg, Fachbereich Physik, Marburg
Metall-Isolator-Übergänge

Institute seminars

K. Kern (02.12.2003)
Max-Planck-Institut für Festkörperforschung, Stuttgart
Molecule by molecule: Building supramolecular nanostructures

A. Winkelmann (03.12.2003)
Friedrich-Schiller-Universität Jena, Institut für Festkörperphysik, Jena
Electron diffraction methods for the analysis of ultrathin silicon carbide films

D. Kim (09.12.2003)
Max-Planck-Institut für Polymerforschung, Mainz
Nanofabrication via thin film block copolymer templates

J. Shen (10.02.2004)
Tongji University, Pohl Institute of Solid State Physics, Shanghai, China
Preparation and application of structure controllable nanoporous materials: Aerogels and aerogel-films

B. Ujfalussy (11.02.2004)
Oak Ridge National Laboratory, Metals and Ceramics Division, Oak Ridge, USA
Ab-initio spin-dynamics in interfaces and surface nanostructures

K. Potzger (17.02.2004)
Hahn-Meitner-Institut, Bereich Strukturforschung, Berlin
Sites with different magnetic properties at Ni surfaces and Ni-Pd-interfaces investigated with PAC spectroscopy

J. Wunderlich (17.02.2004)
Hitachi Cambridge Laboratory, Cavendish Laboratory, Cambridge, United Kingdom
Detection of electric-field induced spin-polarization in a non-magnetic semiconductor heterostructure

H. Riel and W. Riess (01.03.2004)
IBM Research Labs, Zürich, Switzerland
IBM Research Labs in Zürich in general and polymer optoelectronics

B. Ctortecka (16.03.2004)
Garching Innovation GmbH, München
Patenting at an MPI

S. Schlecht (18.03.2004)
Phillipps-Universität Marburg, Fachbereich Chemie, Marburg
Synthesis of compound semiconductor nanoparticles, nanowires and nanotubes from single source precursors

M. Getzlaff (23.03.2004)
Heinrich-Heine-Universität Düsseldorf, Institut für Angewandte Physik, Düsseldorf
Thin film systems: Structure and electronic behavior on the nanometer scale

S. Mangin (21.04.2004)
U.H.P.-Nancy I, Laboratoire de Physique des Matériaux, Vandoeuvre, France
Magnetic configurations at the interface of exchange coupled bilayers

Y. Choi (22.04.2004)
University of Oxford, Department of Materials, Oxford, United Kingdom
Growth and characterisation of epitaxial exchange-biased systems based on chemically disordered Fe-Mn alloy

A. L. Kholkin (26.04.2004)
University of Aveiro, Department of Cermicas & Glass Engineering & Center for Research in Ceramic and Composite Material, Aveiro, Portugal
Nanoscale properties of ferroelectric materials for electromechanical applications

A. Bland (27.04.2004)
University of Cambridge, Cavendish Laboratory, Cambridge, United Kingdom
Chemically selective spin polarisation and spin reorientation in metal nanostructures

R. Spolenak (03.05.2004)
Max-Planck-Institut für Metallforschung, Stuttgart
Contact mechanics of bioinspired fibrillar structures

C. Ross (04.05.2004)
Massachusetts Institute of Technology, Department of Materials Science and Engineering, Cambridge, USA
Properties of magnetic nanostructures - rings, bars and dots

J. Schäfer (11.05.2004)
Universität Augsburg, Institut für Physik, Augsburg
Electronic interactions and phase transitions at surfaces and in low dimensions

H. Braun (14.05.2004)
Eidgenössische Technische Hochschule Zürich, Institut für Theoretische Physik, Zürich, Switzerland
Magnetism at the nanoscale - From superparamagnetism to chiral quantum solitons

E. Umbach (25.05.2004)
Bayerische Julius-Maximilians-Universität Würzburg, Experimentelle Physik II, Würzburg
Growth mechanism and unusual phase transitions of organic molecules on metallic substrates

S. Maekawa (08.06.2004)
Tohoku University, Institute for Materials Research, Sendai, Japan
Non-local spin transport in nanostructures

S. Estreicher (09.06.2004)
Texas Tech University, Physics Department, Lubbock, USA
Vibrational dynamics for impurities in semiconductors

C. L'abbé (09.06.2004)
Katholieke Universiteit Leuven, Department of Physics and Astronomy, Heverlee, Belgium
Nuclear resonant magnetometry

J. Fritz (15.06.2004)
International University Bremen, School of Engineering and Science, Bremen
Nanomechanical and electronic biosensors

J. Ferré (22.06.2004)
Université Paris-Sud, Laboratoire de Physique des Solides, Orsay, France
Magneto-optical study of the magnetization reversal dynamics in ultrathin films and patterned structures

C. Ascheron (24.06.2004)
Springer-Verlag, Heidelberg
Science Citation Index - Use and Abuse. Is the science citation index the ultimate measure for the quality of scientific publications?

T. Mizoguchi (28.06.2004)
Gakushuin University, Faculty of Science, Tokyo, Japan
The study of rust on weathering steel

C. Hess (07.07.2004)
Fritz-Haber Institut der Max-Planck-Gesellschaft, Anorganische Chemie, Berlin
Synthesis, in situ characterization and catalytic application of highly dispersed vanadia supported on mesoporous silica SBA-15

D. Kovalev (08.07.2004)
Technische Universität München, Physik-Department, Garching
Photosensitization of oxygen molecules by silicon nanocrystals

A. Slobodskyy (09.07.2004)
Bayerische Julius-Maximilians-Universität Würzburg, Physikalisches Institut, Würzburg
Magnetic resonance tunneling diodes

B. Yang (12.07.2004)
Kumoh National Institute of Technology, Department of Materials Science and Engineering, Gyeongbuk, Korea
BLT films and integration processes for reliable ferroelectric high density memories

T. Zhang (02.08.2004)
Hong Kong University of Science and Technology, Department of Mechanical Engineering, Kowloon, Hong Kong
Microbridge testing of thin films

B. Heinrich (22.09.2004)
Simon Fraser University, Physics Department, Burnaby, Canada
Spin pumping and two magnon scattering in metallic multilayer films

T. Sorop (24.09.2004)
University of Leiden, Leiden Institute of Physics, Leiden, The Netherlands
Magnetic and superconducting properties of 2D arrays in nanoporous alumina

G. Tatara (27.09.2004)
University of Paris-Sud, Laboratoire de Physique des Solides, Orsay, France
Microscopic theory of current-driven domain wall motion

H. Ebert (27.09.2004)
Ludwig-Maximilians-Universität München, Physikalische Chemie, München
Magnetic and spectroscopic properties of free and supported transition metal clusters

S. Suga (05.10.2004)
Osaka University, Graduate School of Engineering Science, Osaka, Japan
Resonant inelastic X-ray scattering study of Cu-O 1D-chain systems

R. Ramchal (07.10.2004)
Universität Duisburg-Essen, Fakultät für Naturwissenschaften, Duisburg
In situ magnetic domain imaging at the spin-reorientation transition of ultrathin Ni- and Fe/Ni-films

M. Löffler (11.10.2004)
Friedrich-Alexander-Universität Erlangen-Nürnberg, Lehrstuhl für Festkörperphysik, Erlangen
Growth of cobalt silicide films by pulsed laser deposition

Y. Ein-Eli (18.10.2004)
Technion-Israel Institute of Technology, Dept. of Materials Engineering, Haifa, Israel
Silicon negative potential dissolution (NPD) - From polishing to texturing and related mechanism

S. F. Fischer (25.10.2004)
Ruhr-Universität Bochum, Werkstoffe und Nanoelektronik, Bochum
Mode coupling of spatially coincident electron wave guides

E. Vedmedenko (26.10.2004)
Universität Hamburg, Institut für Angewandte Physik, Hamburg
Lattice-dependent anisotropy in the orientation of magnetic domain walls

N. Stefanakis (27.10.2004)
CNRS-CRTBT, Grenoble, France
Interface effects in hybrid structures of carbon nanotubes

Visiting groups

10 visitors attending the event "Physiker in Industrie und Wirtschaft - Ein Tag vor Ort" organized by Deutsche Physikalische Gesellschaft
Halle, January 30, 2004

21 participants from "Deutscher Verein zur Förderung des mathematischen und naturwissenschaftlichen Unterrichts"
Halle, April 07, 2004

10 physics students from Riga (Latvia) as part of their internship at FH Merseburg
Halle, June 09, 2004

2 students attending the 7th Summer School for Girls
Halle, July 08, 2004 (with Martin-Luther-Universität Halle-Wittenberg)

2 "Jugend forscht" award winners
Halle, August 23 – October 01, 2004

60 freshmen from Martin-Luther-Universiät Halle-Wittenberg majoring in physics
Halle, October 06, 2004

21 high-school students from Eisleben majoring in physics
Halle, October 13, 2004

Events for the public at large

Girls' Day - Future Prospects for Girls
Halle, April 22, 2004

3. Lange Nacht der Wissenschaften
Halle, July 02, 2004

University lectures

Bruno, P.
Quantum transport in mesoscopic systems: A panorama
Martin-Luther-Universität Halle-Wittenberg, Halle
winter 03/04

Christiansen, S. H.
Werkstoffe der Photovoltaik: Grundlagen - Materialien und Technologie
Universität Erlangen-Nürnberg, Institut für Werkstoffwissenschaften, Erlangen
winter 03/04

Christiansen, S. H.
Mikrocharakterisierung und Werkstoffverhalten II
Universität Erlangen-Nürnberg, Institut für Werkstoffwissenschaften, Erlangen
summer 04

Christiansen, S. H.
Werkstoffe der Photovoltaik: Grundlagen-Materialien und Technologie
Universität Erlangen-Nürnberg, Institut für Werkstoffwissenschaften, Erlangen
winter 04/05

Ernst, A.
Übungen zur Theoretischen Physik I: Mechanik
Martin-Luther-Universität Halle-Wittenberg, Halle
winter 03/04

Hertel, R., J. Kirschner
Magnetische Nanostrukturen
Martin-Luther-Universität Halle-Wittenberg, Halle
summer 04

Hesse, D., H.-P. Abicht, D. Völtzke
Grundlegende Aspekte der Festkörperchemie und Materialwissenschaften
Martin-Luther-Universität Halle-Wittenberg, Halle
winter 03/04

Hesse, D., H.-P. Abicht, D. Völtzke
Grundlegende Aspekte der Festkörperchemie und Materialwissenschaften
Martin-Luther-Universität Halle-Wittenberg, Halle
winter 04/05

Kuch, W.
Verfahren zur mikroskopischen Abbildung von Festkörperoberflächen
Martin-Luther-Universität Halle-Wittenberg, Halle
summer 04

Meyerheim, H. L., J. Kirschner
Magnetism of Surfaces and Thin Layers
Martin-Luther-Universität Halle-Wittenberg, Halle
winter 03/04

Meyerheim, H. L., J. Kirschner
Magnetism of Surfaces and Thin Films
Martin-Luther-Universität Halle-Wittenberg, Halle
winter 04/05

Sandratskii, L. M.
Theorie der kondensierten Materie - Theoretische Polymerphysik
Martin-Luther-Universität Halle-Wittenberg, Halle
winter 03/04

Werner, P.
Einführung in die Elektronenmikroskopie
Martin-Luther-Universität Halle-Wittenberg, Halle
winter 04/05

Woltersdorf, J.
Graduiertenseminar des FB Physik der Martin-Luther-Universität Halle-Wittenberg für Physik-Doktoranden
Martin-Luther-Universität Halle-Wittenberg, Halle
winter 03/04

Woltersdorf, J.
Graduiertenseminar des FB Physik der Martin-Luther-Universität Halle-Wittenberg für Physik-Doktoranden
Martin-Luther-Universität Halle-Wittenberg, Halle
summer 04

Zacharias, M., M. Hanke
Nanostrukturphysik
Martin-Luther-Universität Halle-Wittenberg, Halle
winter 04/05

Publications and presentations

Journals and books

1 *Akatsu, T.*, *R. Scholz*, and *U. Gösele*.
Dislocation structure in low-angle interfaces between bonded Si (001) wafers.
Journal of Materials Science **39** (9), 3031–3039 (2004).

2 *Akhmetov, V. D.*, *H. Richter*, *W. Seifert*, *O. Lysytskiy*, *R. Wahlich*, *T. Müller*, and *M. Reiche*.
Distribution and properties of oxide precipitates in annealed nitrogen doped 300 mm Si wafers.
European Physical Journal of Applied Physics **27** (1-3), 159–161 (2004).

3 *Alexe, M.* and *U. Gösele*, Eds.
Wafer Bonding - Applications and Technology.
Springer, Berlin, Germany (2004).

4 *Alexe, M.* and *A. Gruverman*, Eds.
Nanoscale Characterisation of Ferroelectric Materials - Scanning Probe Microscopy Approach.
Springer, Berlin, Germany (2004).

5 *Alexe, M.*, *C. Harnagea*, and *D. Hesse*.
Non-conventional micro- and nanopatterning techniques for electroceramics.
Journal of Electroceramics **12**, 69–88 (2004).

6 *Alexe, M.*, *I. Radu*, and *I. Szafraniak*.
Wafer bonding of ferroelectric materials.
In: *Wafer Bonding: Applications and Technology*, Eds. M. Alexe and U. Gösele, 451–472. Springer, Berlin, Germany (2004).

7 *Aliev, F. G.*, *V. K. Dugaev*, and *J. Barnas*.
Localization and interactions in magnetic nanostructures.
In: *Encyclopedia of Nanoscience and Nanotechnology*, Ed. H. S. Nalwa, Vol. 4, 587–605. American Scientific Publishers, Stevenson Ranch (2004).

8 *Andricek, L.*, *G. Lutz*, *M. Reiche*, and *R. H. Richter*.
Processing of ultra-thin silicon sensors for future e^+ e^- linear collider experiments.
IEEE Transactions on Nuclear Science **51**, 1117–1120 (2004).

9 *Bao, D.*, *X. Yao*, *K. Shinozaki*, and *N. Mizutani*.
Growth and electrical properties of Pb(Zr,Ti)O_3 thin films by a chemical solution deposition method using zirconyl heptanoate as zirconium source.
Journal of Crystal Growth **259** (4), 352–357 (2003).

10 *Bartsch, M.*, *A. Tikhonovsky*, and *U. Messerschmidt*.
Plastic deformation of yttria stabilized cubic zirconia single crystals. II. Plastic instabilities.
Physica Status Solidi A **201** (1), 46–58 (2004).

11 *Bartsch, M.*, *Z.-F. Zhang*, *C. Scheu*, *M. Rühle*, and *U. Messerschmidt*.
Fracture parameters of chevron-notched Al_2O_3/Nb sandwich specimens.
Zeitschrift für Metallkunde **95** (9), 779–784 (2004).

12 *Berakdar, J.* and *J. Kirschner*, Eds.
Correlation Spectroscopy of Surfaces, Thin Films, and Nanostructures.
Wiley-VCH, Weinheim, Germany (2004).

13 *Berakdar, J.*
Comment on "Role of the ground state in electron-atom double ionization".
Physical Review Letters **92** (14), 149301/1 (2004).

14 *Breitenstein, O.*, *J. P. Rakotoniaina*, *M. Hejjo Al-Rifai*, and *M. Werner*.
Shunt types in crystalline silicon solar cells.
Progress in Photovoltaics: Research and Applications **12** (7), 529–538 (2004).

15 *Breitenstein, O.*, *J. P. Rakotoniaina*, and *M. Hejjo Al-Rifai*.
Quantitative evaluation of shunts in solar cells by lock-in thermography.
Progress in Photovoltaics: Research and Applications **11** (8), 515–526 (2003).

16 *Brendel, A.*, *C. Popescu*, *C. Leyens*, *J. Woltersdorf*, *E. Pippel*, and *H. Bolt*.
SiC-fibre reinforced copper as heat sink material for fusion applications.
Journal of Nuclear Materials **332**, 804–808 (2004).

17 *Brückner, W.*, *J. Thomas*, *R. Hertel*, *R. Schäfer*, and *C. M. Schneider*.
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14.10.2004.

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Charged particle interaction with magnetic thin films and surfaces.
8th International Workshop on Fast Ion-Atom Collisions (FIAC-04), Debrecen, Hungary.
31.08. - 03.09.2004.

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Bergakademie Freiberg, Freiberg, Germany.
28.05.2004.

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Electronic correlation mapping in nano-size materials.
III International Workshop on Electron Correlations and Materials Properties, Kos, Greece.
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Introduction to spintronics.
IV International School on Magnetism and Synchrotron Radiation, Mittelwihr, France.
10. - 16.10.2004.

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Magnetism of surfaces and nanostructures.
Freie Universität Berlin, Institut für Physik, Berlin, Germany.
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WE-Heraeus-Ferienkurs für Physik "Physik der magnetischen Nanostrukturen", Martin-Luther-Universität Halle-Wittenberg, Halle, Germany.
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Direct wafer bonding: Past, present and future.
Workshop on Wafer Bonding for MEMS Technologies, Halle, Germany.
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Engineered substrates based on wafer bonding for future silicon technology.
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24. - 28.05.2004, Talk.

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Interfaces in nanosize ferroelectrics - structure and impact on properties.
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