Triannual Report 1996 — 1998

Research Group "Scanning Probe Methods"



University of Hamburg Microstructure Advanced Research Center Hamburg and Institute of Applied Physics



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Chapter 1 Preface

This is the second triannual report of the research group "Scanning Probe Methods" at the Institute of Applied Physics and "Microstructure Advanced Research Center Hamburg (MARCH)" of the University of Hamburg. The general goals of MARCH include the fabrication of artificial micro- and nanostructures by complementary techniques (including molecular beam epitaxy, photolithography, interference lithography, electron beam lithography, scanning probe nanofabrication), their investigation by various microscopical and spectroscopical methods, and finally their application in new types of nano- or quantum devices (see also the special brochure about MARCH). In contrast to some other research centers of similar size, the research at MARCH does not exclusively concentrate on a particular class of materials but covers semiconducting, metallic, superconducting, as well as magnetic nanosystems. Special emphasis is also put on investigations of micro- and nano-scale hybrid systems (e.g. semiconductorsuperconductor, semiconductor-ferromagnet, metal-ferromagnet hybrids) and corresponding devices. An outstanding feature of MARCH is undoubtedly its balanced devotion to high-level education, basic research, applied research, technological as well as product developments:

- Education: In the past three years an intensified emphasis on the education of graduate students has been put in the Graduiertenkolleg "Physics of nanostructured solids" funded by the German Science Foundation (current chairman: Prof. Dr. W. Hansen). Our research group is also involved in a second recently established Graduiertenkolleg "Spectroscopy on localized atomic systems" (chairman: Prof. Dr. G. Huber from the Institute of Laser Physics). A series of lectures and symposia entitled "Physics of Micro- and Nanostructures" has been introduced together with the foundation of MARCH which is alternately presented by the academic staff of MARCH, by scientists from other research institutions, or by representatives from industry.
- **Basic research:** In 1997 a new "Sonderforschungsbereich on Quantum Materials" funded by the German Science Foundation (chairman: Prof. Dr. D. Heitmann) was established which greatly supports the basic research activities at MARCH in the areas of low-dimensional electron systems, semiconductor-

superconductor hybrid systems and semiconductor cluster research. The latter research area is of interdisciplinary nature and involves activities at the Institute of Physical Chemistry (Prof. Dr. H. Weller), the Hamburg Synchrotron Radiation Laboratory (HASYLAB), as well as of MARCH. Numerous other basic research projects, e.g. in the area of nanomagnetism and spin electronics, are funded by the German Science Foundation, the VW Foundation as well as the European Community.

• Applied research, technological and product developments: To complement the basic research activities by applied research projects and technological developments leading to new products in the area of micro- and nanotechnology is another major goal of MARCH. The recently established national "Center of Competence in Nano-scale analysis" (Coordinators: Prof. Dr. H. Fuchs, University of Münster, Prof. Dr. H. Gaub, University of Munich and Prof. Dr. R. Wiesendanger, MARCH and University of Hamburg) funded by the Federal Ministry for Education, Science, Research and Technology (BMBF) greatly stimulates product oriented research and developments in the field of nano-instrumentation. A powerful network of about 40 members from universities, research centers, small and medium-size companies as well as big companies has been created which will effectively contribute to convert basic research results into competitive 'high-tech' products on a short time scale. Outstanding examples for new product developments at MARCH are the first ultra-high vacuum (UHV) compatible, low-temperature atomic force microscope with true atomic resolution, the first UHV-compatible magnetic force microscope with single magnetic layer sensitivity, the spin-polarized version of the scanning tunneling microscope for nano-scale magnetic domain imaging and a high-sensitivity scanning capacitance sensor for high-resolution dopant profiling. A new service center for nano-scale analysis affiliated at MARCH is currently being established which will help to further stimulate the technology transfer between MARCH and industry. The activities of MARCH in these applied and technological areas are increasingly supported directly by industrial companies, including Beiersdorf, Digital Instruments, Exabyte Magnetics, Oxford Instruments, Siemens, and Surface Imaging Systems.

Another major event in the past time period has been the organisation of the 9th International Conference on Scanning Tunneling Microscopy/Spectroscopy and Related Techniques (STM'97) which was held in the Congress Center Hamburg (CCH) during July 20 - 25, 1997. The STM conference continued to grow in size and scope, indicated by the record number of 826 scientific presentations and the record of 1033 participants representing 38 nations from 5 continents. Another record achieved was the participation of 31 companies in the industrial exhibition, indicating that STM-based science and research has increasing impact on the technological developments in 'high-tech' areas. The great participation of young students and researchers in the STM'97 conference provides confidence that this research field will continue to be of outstanding vitality, excitement, and full of visions for the future of science and technology.

This research report provides a good opportunity to thank all funding agencies as well as industrial companies for their support of our activities in nano-science research and organization. In particular we would like to thank the Ministry of Science and Research of the City of Hamburg for its strong devotion to establish MARCH as an internationally recognized Center of Excellence in nano-science and technology. I would also like to take the opportunity to thank all members of the research group "Scanning Probe Methods" for their outstanding efforts and achievements during the past three years. Many thanks also to Oswald Pietzsch and Hendrik Hölscher for acting as 'editors' of this triannual report. Finally, we gratefully acknowledge the excellent support by our central mechanical and electronical workshops, as well as by our secretaries and administration staff of our physics department.

Hamburg, December 1998

Prof. Dr. Roland Wiesendanger

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Chapter 3 Research Activities 1996-98

3.1 Overview

R. Wiesendanger

Our research activities are concentrated on nanometer-scale probe methods. In particular, emphasis is put on the investigation of the fundamental relationship between nanostructure and nanophysical properties. Scanning probe methods (SPM) are ideally suited for such investigations because they provide high spatial resolution combined with spectroscopical capabilities. By choosing an appropriate type of interaction between probe tip and sample, almost any kind of nanophysical property can be studied by SPM. We apply SPM to various classes of materials, including metals, semiconductors, insulators, superconductors, magnetic materials, as well as organic thin films and materials.

To be able to make significant contributions to this rapidly developing field, a major part of our activities is devoted to new developments or further improvements on SPM instrumentation, including the development of new positioning devices with nanometer-scale accuracy, the development of new types of sensors, or the development of dedicated SPM instruments which can operate under extreme conditions (e.g. UHV, low temperatures and high magnetic fields). Special emphasis is also put on the development of new experimental methods based on the local probe geometry, which usually requires adjustments of the hardware and software for SPM data acquisition. These developments are often made in close collaboration with European companies specialized in SPM instrumentation.

In the following, a brief summary of the highlights of our research activities in the time period of 1996 - 1998 is provided which will be decribed in more detail in the subsequent chapters. This may help to guide you directly to the chapters of your specific interest. To keep the extent of this report at a reasonable level we will, of course, only provide short summaries of the results of our research projects. For further information, please do not hesitate to contact us. We will be glad to provide reprints of publications on specific topics.

Nanomagnetic studies

Our research activities in this field concentrate on the correlation between nanostructural, local electronic and local magnetic properties of magnetic thin films, multilayer systems, as well as laterally structured thin films and multilayers ('magnetic dot arrays'). The experimental methods for these investigations include variabletemperature spin-resolved scanning tunneling spectroscopy and magnetic force microscopy (both under ambient and UHV conditions) in combination with spin-resolved photoelectron spectroscopy, Hall-probe magnetometry (in cooperation with the semiconductor research group of Prof. D. Heitmann) and SQUID magnetometry (in cooperation with the low-temperature physics group of Prof. J. Kötzler).

A recent outstanding achievement was the first demonstration of spin-polarized vacuum tunneling into exchange-split surface states which allowed to image the nanomagnetic domain structure of Gd thin films as well as to clearly separate between structural, electronic and magnetic information. The quantitative values of the measured local spin polarization were found to be in excellent agreement with results from spin-resolved photoemission and inverse photoemission spectroscopy. Non-destructive magnetic imaging by SP-STS could be achieved by using magnetically coated STM tips with a thickness of the magnetic coating well below 100 nm in order to reduce the tip's magnetic stray field. The technology of in-situ tip preparation for magnetic studies has been introduced earlier when we have realized the first UHV-compatible magnetic force microscope (MFM) with single magnetic layer sensitivity. The UHV-MFM instrument has recently been applied to study the reorientation transition of magnetic anisotropy in ultra-thin Co/Au(111) films at high spatial resolution. In addition, first steps towards a combination of UHV-MFM, being sensitive to the magnetic stray field distribution, and spin-polarized tunneling spectroscopy, being sensitive to the spin-resolved local electronic structure, were made. MFM has also been applied to study local magnetic switching properties of single-domain particles as a function of shape, size and inter-particle distance in magnetic dot arrays being prepared either by interference lithography or electron-beam lithography (in cooperation with Prof. D. Heitmann, Hamburg and Prof. D. Weiss, Regensburg). These experimental investigations have been complemented by theoretical simulations based on micromagnetic model calculations. Finally, MFM has been used extensively for quality control in the field of magnetic data storage media.

Nanoelectronic studies

In this field our research activities are focussed on local tunneling spectroscopic studies of electronic states in III-V semiconductors at low temperatures and in high magnetic fields. Recently, we have studied Landau quantization of electronic states in InAs in magnetic fields up to 7 T. The high quality of our spectroscopic data allowed to determine the energy-dependence of the effective electron mass in InAs up to 250 meV. Spatially resolved spectroscopic studies of the Landau quantization led to the observation of spatial potential fluctuations on the order of a few mV induced by inhomogeneities in the dopant density distribution. In a second study, we concentrated on the energy- and magnetic-field dependence of electronic scattering states at dopant sites. Quantitative analysis of the ring-like patterns of these scattering states as a function of energy allowed to determine the depth of the dopants below the surface. Surprisingly, dopants located up to several ten nanometers below the surface were found to influence the local density-of-states distribution at the surface as probed by STS. In the presence of an external magnetic field, the electrons cannot scatter into states with arbitrary wave vectors but only into states with wave vectors falling onto the Landau tubes in k-space. This could be made visible for the first time by calculating 2D Fourier transforms of real-space STS data.

In a third study, we have measured the magnetic-field dispersion of tip-induced quantum dot states in InAs which allowed a quantitative determination of tip-induced band-bending effects at III-V semiconductor surfaces. The experimental results were found to be in excellent agreement with Hartree-Fock calculations taking electronelectron correlation effects into account.

This work on spectroscopic STS studies of electronic states in InAs is performed in close collaboration with the research groups of Prof. U. Merkt, Prof. Dr. D. Heitmann (both at MARCH) and Prof. V. Gudmundsson (Reijkjavik).

A second activity of our research group in the field of nanoelectronic studies is devoted to Si-based devices being probed by scanning capacitance microscopy (SCM) and spectroscopy.

Recently, we have demonstrated dopant profiling by SCM on the nanometer scale with great potential for quality control and failure analysis in the semiconductor industry. The experimental investigations are complemented by theoretical device simulations which are expected to lead to a quantitative analysis of dopant concentration profiles by SCM-based methods. Scanning capacitance spectroscopy will be developed further for the investigation of electronic states in semiconductor quantum structures at low temperatures. This work is done in close collaboration with the research group of Prof. W. Hansen at MARCH.

Nanomechanical studies

Our research activities in this field concentrate on quantitative investigations of frictional phenomena on the nanometer scale as well as on the quantitative determination of interaction potentials based on low-temperature scanning force microscopy (SFM) data. Recently, we have demonstrated the first quantitative and reproducible determination of the microscopic frictional coefficient by using spherical-symmetric probe tips which allowed to analyse and fit the experimental data by an extended Hertzian theory. This way we could derive a microscopic coefficient of friction being independent of the size of the probe tip or the area of the tip-sample contact. We have further shown by comparison between experimental SFM data and theoretical simulations based on solving the equations of motion that atomic force images obtained in the contact mode of operation are strongly influenced by lateral or frictional forces. Furthermore, the relatively large contact area prevents measurements with 'true atomic resolution'. We have therefore developed a UHV-compatible low-temperature atomic force microscope which can operate in the dynamic non-contact mode. This instrument allowed to achieve 'true atomic resolution' on InAs(110) as well as on van der Waals surfaces such as graphite(0001) or xenon(111) films. Due to the extremely high mechanical stability of this system it was possible to obtain high quality force spectroscopy data which allowed to extract information about tip-sample interaction potentials. Furthermore, we developed a theory for explaining the atomic resolution capability of the dynamic SFM operation mode. The UHV low-temperature SFM system is currently applied to study charged dopants in III-V semiconductors based on the electrostatic force interaction as well as for atomic-resolution studies of condensed noble gas crystals.

In summary, our low-temperature SPM activities nicely demonstrate that this technique is becoming of increasing importance in modern solid state research besides the more specialized field of surface physics in which STM and related techniques have proven to be invaluable already a long time ago. We can now resolve charge and spin states as well as determine effective masses of electrons on a local scale.

What about next? The spatially resolved study of elementary excitations, such as phonons or magnons, in the presence of defects such as steps or impurities, constitutes a major challenge for the future. To address such issues will be in the focus of our future research.

3.2 Nanomagnetism

3.2.1 On the Development of Spin–Polarized Scanning Tunneling Spectroscopy

M. Bode, M. Getzlaff, R. Pascal, H. Tödter, and Ch. Zarnitz

Introduction

The spectroscopic capabilities of the scanning tunneling microscope (STM) [1] open up the fascinating possibility of correlating the local structural and electronic properties on the atomic scale with magnetic properties. In the past many attempts have been made to gain magnetic information by making the STM sensitive to the spin of the tunneling electrons. Basically two different concepts have been used to achieve spin– polarized vacuum tunneling:

(1) By using magnetic STM probe tips the spin-value effect [2] can be exploited which relies on the fact that the tunneling conductance between two ferromagnetic electrodes separated by an insulating barrier depends on whether the magnetic moments are directed parallel or antiparallel. This effect has been extensively studied in planar tunneling junctions [3–5] and has been used in STM to probe the topological antiferromagnetic order of a Cr(001) surface by means of a CrO_2 -tip [6].

(2) Optically pumped GaAs tips or samples enable spin-polarized vacuum tunneling to be observed [7] and this technique has been applied to image the magnetic domain structure of thin Co films [8].

All of the experiments performed to date are limited by the need to separate topographic, electronic, and magnetic information in the case of magnetic probe tips, and to eliminate thermal or film thickness induced effects in the case of semiconducting tips.

Our experimental approach to overcome these difficulties is based on tunneling into the well-known surface state of Gd(0001) which is exchange-split into a filled majority and an empty minority spin-contribution [9, 10]. In analogy to the low-temperature experiments performed with ferromagnet-insulator-superconductor planar tunneling junctions [11, 12] where the quasiparticle density of states of superconducting aluminum is split by a magnetic field into spin-up and spin-down parts, we use two spin-polarized electronic states with opposite polarization to probe the magnetic orientation of the sample relative to the tip. In order to prevent any misinterpretation, prior to magnetic contrast studies we have performed a detailed investigation on the thickness-dependence of structural and electronic properties of Gd thin films grown on W(110). Our studies revealed a rich variety of different superstructures and belonging electronic properties already at submonolayer coverage which are described on page 13*ff*. For coverages above one monolayer we used three different preparation procedures that result in different film morphologies as illustrated on page 18 ff. The Gd(0001) surface state appears as a double-peak structure in the tunneling spectra that quenches upon adsorption of hydrogen (page 19 ff). On pages 22 ff we will describe in how far the tunneling spectra depend on the film thickness. Only then we will present magnetic contrast studies performed with Fe-covered probe tips (page 32 ff). In fact, we were able to demonstrate spin-polarized tunneling by measuring the asymmetry of the differential tunneling conductivity at bias voltages corresponding to the energetic positions of the two spin-contributions of the exchange-split surface state in an external magnetic field. This enables the electronic and magnetic structure information to be clearly separated. By mapping the spatial variation of the asymmetry parameter we have been able to observe the nanomagnetic domain structure of Gd(0001) ultra thin films with a spatial resolution below 20 nm.

Results and Discussion

Gd on W(110) at submonolayer coverage: structural and local electronic properties $% \left(\frac{1}{2} \right) = 0$

Although Gd(0001) films on W(110) have been intensively studied in the past for the coverage regime of one up to several ML,very little is known about the submonolayer coverage regime. The morphology of the films was first studied by low energy electron diffraction (LEED) and Auger electron spectroscopy (AES) [13,14]. These techniques are known to average over at least several tenths of a square millimeter. A more detailed view on the film morphology can be achieved by scanning tunneling microscopy (STM) which yields a real space image of the topography on the relevant scale of several 10–100 nm. Recently, a STM study on the growth of Gd on W(110) in the coverage range of 1–20 ML with the focus on 11 ML was published [15]. In this first



Figure 3.1: Constant-current STM topograph (scan-range: 60 nm × 60 nm) of 0.5 ML Gd on W(110) annealed at 710 K. Different superstructures are observed at different locations: (a): (8×2) , (b): (7×2) , (c): (6×2) , (d): (5×2) , and (e): $c(5 \times 3)$. The circle marks a dislocation. Tunneling current I = 1 nA, sample bias voltage U = -0.8 V.

part of our contribution we report on a STM-study of the growth of Gd/W(110) at submonolayer coverage. For the benefit of comparability we chose the same annealing temperatures as Tober et al. [15], i.e. 530 K and 710 K.

Figure 3.1 shows a topographic STM image of a 60 nm × 60 nm area of a nominally 0.5 ML Gd film on W(110) annealed at 710 K for 10 minutes. Five different one-dimensional surface structures (labeled a-e) appear in the STM-image which correspond to (8×2) (a), (7×2) (b), (6×2) (c), (5×2) (d) and $c(5 \times 3)$ (e) superstructures of the thin Gd film. These superstructures are characterised by $(n \times 2)$ LEED-patterns with n = 10, 8, 7, 6, 5.

Figure 3.2 shows line sections along the [001]-direction, i.e. perperdicular to the chains, which were measured at negative bias voltage (U = -0.7 - 0.8 V). Obviously, the apparent corrugation decreases with decreasing n. At this particular bias voltage it is up to 1.2 Å for the (10×2) -structure but it did not exceed 0.1 Å for the (5×2) -structure . However, the measured corrugation of $(n \times 2)$ -structures with $n \geq 7$ strongly depends on the applied bias voltage as shown in Fig. 3.2(b) for the (8×2) -structure. While we have measured 0.45 ± 0.02 Å at U = -0.8 V the corrugation nearly vanishes at U = 0.8 V $(0.03 \pm 0.01$ Å) and inverts for higher positive bias voltages.

To our knowledge a first indication of chain formation was found by Weller and Alvarado [13]. They report on 8th–order streaks in their LEED data along the [001]–



Figure 3.2: (a) Line-sections drawn along the [001]-direction, i.e. perpendicular to the chains. The data for all $(n \times 2)$ -superstructures except for n = 10 (I = 1 nA, U = -0.7 V) were extracted from the same STM-image (I = 1 nA, U = -0.8 V). (b) Bias voltage dependent corrugation of the (8×2) -superstructure. Close to U = 0.8 V the apparent corrugation inverts.

direction at a coverage $\Theta \approx 0.3$ ML. Kołaczkiewicz and Bauer [14] concentrated on the growth at submonolayer coverage at 300 K and 1200 K. At very low coverages ($\Theta \approx 0.15$ ML) they reported on stripes in the LEED-pattern along the [001]direction. With increasing coverage the streaks soon resolve into spots of $(n \times 2)$ patterns with n = 10, 7, 6, 5 which are best resolved at $\Theta = 0.2, 0.28, 0.33$ and 0.4 ML, respectively [14]. Kołaczkiewicz and Bauer explained these LEED-patterns by Gdrow formation along the [110]-direction which is driven by charge transfer from the electropositive adsorbate to the substrate [14]. This results in a repulsive Coulombinteraction between adjacent gadolinium atoms. Since this interaction exhibits no intrinsic anisotropy the anisotropy of the W(110)-substrate is of substantial importance for the formation of Gd chains. To explain the observations the repulsive Coulombinteraction must dominate along the [001]-direction while it is overcompensated by an attractive interaction along the [110]-direction.

Besides the (8×2) -structure which—to the best of our knowledge—has not been



Figure 3.3: Atomic structure models of the submonolayer superstructures formed by Gd/W(110) adopted from Ref. [14].

described before our LEED and STM-results are in excellent agreement with an earlier work by Kołaczkiewicz and Bauer [14], who proposed a structure model which we have extended to the (8×2) -structure [Fig. 3.3]. Along the [001]-direction the gadolinium adsorbates occupy alternating on top and non-top sites. The latter are fourfold hollow sites for the (10×2) -, (8×2) -, and (6×2) -structure and bridge sites in the case of the (7×2) - and the (5×2) -structure. Thus chains and troughs in our STM-images represent Gd atoms adsorbed at on top and hollow sites, respectively, The measured periodicities along the [001]-direction and the structure-model-based periodicities (in brackets) are: (10×2) : 15.4 Å ±0.5 Å (15.80 Å), (8×2) : 12.6 Å ±0.1 Å (12.64 Å), (7×2) : 10.7 Å ±0.5 Å (11.06 Å), (6×2) : 9.2 Å ±0.3 Å (9.48 Å), and (5×2) : 7.8 Å ±0.3 Å (7.90 Å).

As already shown in Fig. 3.2(a) the measured corrugation decreases with increasing Gd density. However, corrugations up to 1.2 Å as measured with the STM cannot be explained solely by the different adsorption topography since the corrugation of dense–packed metal surfaces is typically below 0.1 Å. In this context one should call in mind that constant current STM-images represent contour maps of constant local density of states (LDOS). Therefore, we have performed scanning tunneling spectroscopy (STS) measurements by simultaneously recording topographic data and the normalized differential tunneling conductance (dI/dU)/(I/U) as a function of the applied sample bias voltage U. In Fig. 3.4 we have plotted the tunneling spectra (a–e) corresponding to the superstructures (a–e) of Fig. 3.1. Since the normalization procedure leads to



Figure 3.4: Normalized tunneling spectra as measured for the five different superstructures already observed in the STM image of Fig. 3.1. Black curves correspond to tunneling sites above the maxima of the corrugation and the grey curves to tunneling sites above the minima, as seen at -0.8 V. problems around zero bias the region from -0.2 V to 0.2 V has been omitted. For the (8×2) and the (7×2) superstructures we made a distinction of tunneling sites above the Gd chains representing on-top (bridge) positions, respectively, and chains representing hollow positions. This was possible due to the relatively large inter-chain distance for these structures. For the more dense structures we found no significant difference between these two sites. By comparing the tunneling spectra (a)-(e) obtained for the submonolayer coverage regime strong differences are observed, particularly for the empty states for which peak positions as well as peak intensities show a strong dependence on surface structure. A systematic trend in peak positions and intensities seems to be present by going from the (8×2) superstructure with a relatively large inter-chain distance to the (5×2) structure with the smallest inter-chain distance. In spectrum (a) we observe a strong peak A at $U \approx 1$ eV. The strong peak in spectrum (b), labeled A, may be interpreted as the peak A of spectrum (a) shifted towards E_F due to the decreased inter-chain distance. This interpretation is supported by the observation that this trend continues for peak A as well as for a new peak B by going to the next closest structure, i. e. (6×2) in spectrum (c). Peak B also shifts towards E_F by going from the (6 \times 2) to the (5 \times 2). The spectrum (e) of the pseudohexagonal $c(5 \times 3)$ structure does not follow this trend, indicating that it exhibits different local electronic properties.

Additional information about the local electronic structure of the Gd films on W(110) in the submonolayer regime was gained from spatially resolved measurements



Figure 3.5: (a) Spatially resolved map (80 nm \times 60 nm) of the tunneling barrier height dI/dz reflecting spatial variations in the local surface work function (tunneling parameters: U = -0.7 V, I = 1 nA). The nomenclature used to mark the superstructures is the same as in Fig. 3.1. (b) Line-sections of the dI/dz-signal as measured above different superstructures. (c) The lowest barrier height is measured above the (5 \times 2) and (6 \times 2) superstructures, in agreement with work function measurements [14].

of the local tunneling barrier height which reflect spatial inhomogeneities in the local surface work function. Figure 3.5(a) shows a barrier height map of 0.5 ML Gd on W(110) which exhibits the same superstructures (a-e) already known from the STM image of Fig. 3.1. Additionally, a small surface area (f) with the structure of a closepacked first ML is observed. Six section lines of the different superstructures in the dI/dz-map are displayed in Fig. 3.5(a). Obviously, the tunneling barrier height and, therefore, the local work function decreases with decreasing inter-chain distance (a-d) by going from the (8×2) to the (5×2) superstructure. However, the next dense structure, i.e. the $c(5 \times 3)$, shows a distinctly higher work function compared to the (5×2) structure. As presented in Fig. 3.5(c) this STM-based result on the nanometer scale is in excellent agreement with work function measurements reported earlier by Kołaczkiewicz and Bauer [14]. They observed the work function minimum to be associated with the maximum development of the (5×2) and (6×2) LEED pattern of a W(110) sample onto which Gd was continuously evaporated. While spatial averaging of different superstructures is inherent by using this method, the STM offers the possibility to directly observe the change of the local work function for each individual superstructure.

Preparation procedures and morphologies of Gd(0001)-films on W(110)

Previous to any investigation of electronic or magnetic properties of Gd(0001)/W(110)we have systematically examined the influence of different preparation procedures on the the Gd(0001) film morphology. Figure 3.2 shows constant current images of the three typical sample morphologies used in this study. Our preparation procedure is based on an earlier publication of Aspelmeier, Gebhardter, and Baberschke [16]. The authors described phenomenologically the so-called "critical curve", indicating the thickness dependent annealing temperature above which the transition from flat film growth to island formation occurs. Consistently with this work evaporation of $45 \pm$ 5 ML Gd on the W(110)-substrate held at 293 K and subsequent annealing at 710 K for two minutes leads to smooth Gd(0001)-films [Fig. 3.2a)]. The atomically flat terraces are 10 nm $- 1 \ \mu m$ wide. Although these films gave a sharp (1×1) LEED-pattern and the surface state is well-pronounced in photoemission spectra some line defects are visible on the surface [see arrow in Fig. 3.2a)]. In contrast, island formation could be observed at lower coverage of approximately 10 ML even if the substrate temperature did not exceed 530 K [Fig. 3.2b)]. These islands exhibit a local coverage of at least 4 ML, are atomically flat on top and will be described in more detail below. Between the islands the W(110)-substrate is covered by a wetting layer of Gd which is hexagonal but heavily distorted. Two different models of the atomic structure of this so-called (14×7) -structure have been described by other authors [14, 15]. Recent atomically resolved STM-data could not support these models [17]. Subsequent deposition of half a monolayer at room temperature on a sample similar to Fig. 3.2b) results in samples with a large variety of local coverages. Due to the reduced mobility, nucleation of second and third layer patches occur on the strained monolayer of Gd/W(110),





Figure 3.6: Constant current topographs of typical sample morphologies used in this study: a) Smooth Gd(0001)-film grown on W(110) by evaporation of ≈ 45 ML at a substrate temperature of 293 K and subsequent annealing at 710 K for 2 min. The arrow points to a line defect on the surface. **b**) Gd(0001)-islands prepared by deposition of 10 ML on the substrate held at 530 K. Between the islands the substrate is covered by a distorted hexagonal monolayer. c) Nucleation of second and third monolayer patches of Gd(0001) can be observed if 0.5 ML are deposited on a sample similar to b) after cooldown to 293 K. Homoepitaxial growth occurs on the thick Gd-islands. Every scale bar corresponds to 200 nm

and homoepitaxial growth of Gd can be observed on Gd(0001)/W(110) [Fig. 3.2c)]. Such a sample is favorable for a spatially resolving technique like STM/STS since it enables one to measure the surface electronic properties of all apparent coverages simultaneously in a single scan.

The Gd(0001) surface state

In contrast to the commonly known itinerant ferromagnets Fe, Co, and Ni for which the conduction electrons carry the magnetic moment *and* are responsible for the exchange coupling, the rare earth metals are so-called local-moment systems, i.e. the magnetic moment is localized in the atomic 4f-shell. Gadolinium (Gd) with the electron configuration $[Xe]4f^{7}5d^{1}6s^{2}$ has often been regarded as a prototype rare-earth metal since it exhibits a half filled 4f-shell which carries —according to Hund's rule the maximal possible f-shell magnetic moment of $7\mu_{\rm B}$. Below the Curie-temperature of 292.5 K Gd is ferromagnetic. The atomic 4f-moments are exchange coupled via the Ruderman-Kittel-Kasuya-Yosida (RKKY) interaction mediated by the 5d-like conduction electrons which also give a small contribution to the total magnetic moment of about $0.6\mu_{\rm B}$. However, the detailed relationship between the electronic structure and magnetism of the bulk and —even more drastically— the surface have been discussed controversally. The discussion was triggered off by a combined magneto-optic Kerr effect (MOKE) and spin-polarized low-energy-electron diffraction (SPLEED) study on the magnetism of the Gd(0001)-surface by D. Weller *et al.* [18]. The authors claimed to observe an enhanced surface Curie-temperature $T_{\rm CS}$ and a nonferromagnetic, possibly antiferromagnetic coupling between the bulk and the surface layer.¹ These extraordinary surface magnetic properties remained unexplained until Freeman and co-workers [20] proposed the existence of a highly localized d_{z^2} -like surface state. According to this calculation the surface state is exchange split in an occupied majority and an empty minority branch. This difference in occupation was made responsible for an enhanced surface magnetic moment and $T_{\rm CS}$. Indeed, the existence of the occupied majority part of the d_{z^2} -like surface state was soon confirmed experimentally by angle-resolved photoemission spectroscopy (ARPES) [21]. Later on also the empty minority part of the surface state was found by inverse photoemission (IPE) [22]. Further structural and magnetic predictions of this calculation, however, turned out to be incorrect, namely the expansion of the outermost interlayer spacing of Gd(0001) by $\approx 6\%$ and an antiferromagnetic coupling between bulk and surface. Instead, a quantitative LEED-study showed that the outermost interlayer spacing is contracted by $3.5 \pm 1.0\%$ [23]. Furthermore, it is nowadays opinion that bulk and surface of Gd(0001) couple FM instead of AFM [10, 24, 25]. As described in detail by Bylander and Kleinman [26] the discrepancy between theoretical and experimental data might be caused by a general underestimation of exchange, correlation, and magnetic energy within the FLAPW method which cannot be corrected in the case of Gd.

It was one basic question in the framework of this study whether it is possible to observe the Gd(0001) surface state by means of STS. Figure 3.7(a) shows photoemission (PE) and inverse photoemission (IPE) data of Gd(0001) published by Weschke *et al.* [27]. While the PE measurement shows the occupied part of the surface state at a binding energy $E_{\rm bin} \approx 100$ meV the empty part appears in the IPE spectrum at $E_{\rm bin} \approx -250$ meV. It is an advantage of STS to detect the contour of the local density of states (LDOS) on both sides of the Fermi–level by tuning the bias voltage from neg-

¹Since single crystalline surfaces of rare–earth metals are extremely difficult to clean almost all experiments have been performed on thin Gd–films with the epitaxial relationship (0001)Gd \parallel (110)W and [11 $\overline{2}0$]Gd \parallel [1 $\overline{1}0$]W [19]. It was found that Gd(0001)/W(110) is fully relaxed at a coverage of approximately 35 monolayers (1 ML $\simeq 2.89$ Å).



Figure 3.7: a) Photoemission (left) and inverse photoemission (right) spectra of Gd(0001) (from Weschke et al. [27]). The occupied part of the surface state appears in the PE spectrum (binding energy $\approx 100 \text{ meV}$) while the empty part is weakly visible in the IPE spectrum (binding energy $\approx 250 \text{ meV}$). b) Tunneling spectrum measured on a sample similar to Fig. 3.2b) above a Gd(0001)-island and above the first distorted ML at 293 K showing both spin-parts of the surface state. Spatially resolved data at sample bias values indicated by small arrows (c-e) will be shown in Fig. 3.9.

ative to positive bias or vice versa. Indeed, the dI/dU-spectrum measured with the tip positioned above a Gd(0001)-island [topography similar to Fig. 3.2b)] exhibits two distinct maxima at a sample bias U = -0.1 V and U = 0.3 V, being in good agreement to the binding energies for the occupied and the empty part of the surface state as determined by PE and IPE [cf. Fig. 3.7a)], respectively. As described above between the Gd-islands a closed hexagonal but heavily distorted monolayer of Gd covers the W(110) substrate. In contrast to the Gd(0001)-islands the spectra measured above the monolayer exhibit only one asymmetric peak at a positive sample bias U = 0.3 V, i.e. in the empty part of the Gd(0001) surface state at room temperature we will show below that this peak does *not* represent a surface state as claimed in an earlier IPE-study [22].

It is a typical property of a surface state that it vanishes upon exposure of the clean surface to even small amounts of adsorbates. We have checked that both peaks in the STS-spectra that are characteristic for the surface state can be quenched by hydrogen, oxygen, or carbon monoxide down to exposures of approximately 1 L (1 Langmuir $= 1 \times 10^{-6}$ torr·s). Figure 3.8(a) shows the topography of Gd(0001)-islands after exposure to 0.2 L hydrogen. While the surface of island A is homogeneous two different



Figure 3.8: **a)** Topography of Gd(0001)-islands after exposure to hydrogen measured at a sample bias U = -0.3 V. The surface of island A is homogeneous. Some protruding areas (arrow C) can be recognized on island B with an apparent height being 1.4-1.7 Å higher than the surrounding island surface (arrow D). **b)** Tunneling spectra reveal that peaks being characteristic for the surface state are quenched at locations with a reduced apparent height (D). In order to keep the tunneling current constant the tip has to approach closer towards the sample wherever hydrogen has been adsorbed than on clean Gd(0001), consequently clean Gd appears protruding.

apparent heights can be recognized on island B. Some protrusions (e.g. arrow C) appear 1.4–1.7 Å higher than the surrounding island surface (arrow D). Typical tunneling spectra measured above both parts of this inhomogeneous Gd–island are shown in Fig. 3.8(b). The double peak structure being characteristic for the clean Gd(0001) surface state can be recognized in the spectra measured above the protruding areas (arrow C). In contrast, the surrounding island surface exhibits a much lower dI/dU–signal at low positive and negative sample bias indicating a strongly reduced density of states around the Fermi–level caused by the local adsorption of hydrogen. A closer inspection of the Gd–islands reveals that most of them remain clean, e.g. island A, while some are contaminated by hydrogen as island B. This is a consequence of the fact that adsorption preferably starts at surface imperfections while defect–free islands remain unaffected at low hydrogen partial pressure.

Thickness and morphology dependent electronic properties of Gd(0001) on W(110)

We have seen that the surface state appears in the dI/dU-spectra as a double peak structure which can be quenched by the local adsorption of hydrogen. In the following, we would like to discuss whether or not a dependence of the differential conductivity dI/dU on the island thickness exists. Figure 3.9(a) shows the topograph of a sample prepared by deposition of 5 ML Gd/W(110) at 530 K resulting in Stranski–Krastanov–growth, i.e. island formation. Two islands can be recognized. Both islands are atomically flat on top. Below the islands the substrate exhibits several monoatomic steps. Therefore the coverage increases by going from the right to the left island edge as schematically indicated in the line–section [Fig. 3.9b)]. We found local coverages of 7 ML $\leq \Theta_{\text{loc}} \leq 19$ ML and 4 ML $\leq \Theta_{\text{loc}} \leq 22$ ML for the island in the upper and lower part of the image, respectively. Since the interlayer spacing $d_{\text{Gd}(0001)} = 2.89$ Å exceeds $d_{\text{W}(110)} = 2.23$ Å the substrate miscut is overcompensated and the Gd–islands



Figure 3.9: **a)** Constant-current STM topograph of nominally 5 ML Gd(0001)/W(110). The substrate was held at 530 K during evaporation resulting in Stranski-Krastanov-growth. **b)** Line section extracted from a). Since the substrate exhibits steps below the atomically flat islands the local coverage of the Gd-island increases from 7 ML up to 19 ML by going from the right to the left. Maps of the differential conductivity dI/dU measured at different sample bias are shown in **c)** U = +0.8 V, **d)** U = +0.47 V, and **e)** U = -0.1 V. The dI/dU-signal and therefore the local density of states does not depend on the local coverage.

exhibit a slope although they are atomically flat on top.

Simultaneously with the topography we have measured a dI/dU-spectrum at every pixel of the scan. During the measurement the sample was held at T = 293 K. Figures 3.9c)-e) show maps of the differential conductivity dI/dU for different sample bias: c) U = +0.8 V, d) U = +0.47 V, and e) U = -0.1 V. These sample bias values have been marked by arrows at the bottom axis of Fig. 3.7b). The differential conductivity is grey coded, i.e. the higher the local dI/dU-signal the brighter a location appears. At a sample bias U = +0.8 V the tunneling current is dominated by electrons which tunnel from the tip into unoccupied sample states with a binding energy of +0.8 eV. Comparison with the topographic data of Fig. 3.9a) reveals that at this particular binding energy the differential conductivity above the Gd monolayer is higher than above any island. Beside a few small bright spots the dI/dU-signal at U = +0.8 V measured above the Gd island is uniform and therefore independent of the local coverage in the range $4 \leq \Theta_{\rm loc} \leq 22.^2$ In the dI/dU-maps we never found a contrast on the Gd island at any sample bias (cf. Fig. 3.9c)-e) in the voltage range under study (-0.6 V $\leq U \leq +0.9$ V). At U = +0.47 V the contrast between the island surface and the monolayer vanishes. Comparison with the spectra of Fig. 3.9b) reveals that at this sample bias the dI/dU-signal of the Gd monolayer is equal to Gd(0001) islands. The contrast inverts if the sample bias is further reduced. For instance Fig. 3.9e) shows a map of the dI/dU-signal at U = -0.1 V, i.e. close to the energetical position of the occupied surface state.

Until now we have shown that the electronic structure as it manifests in the dI/dUspectra of clean Gd(0001)-islands, i.e. the energetical position of the surface state and its intensity, does not dependent on the local coverage $\Theta_{\text{loc}} \geq 4$ ML. We know, however, that the first monolayer of Gd/W(110) does not exhibit the surface state [cf. Fig. 3.7b)]. Consequently, the question arises at which critical thickness the surface state appears in the tunneling spectra? To unravel this problem we have performed tunneling spectroscopy on a sample with a morphology similar to Fig. 3.2c) prepared by room temperature deposition of 0.5 ML Gd on a Stranski-Krastanovfilm of Gd/W(110). Figure 3.10a) shows a Gd(0001) island named A. On top of the island surface homoepitaxial growth of triangular shaped Gd-islands (e.g. B and C) can be observed. Second (D) and third (E) monolayer patches have nucleated on the first monolayer. The smallest patches exhibit an island area of approximately 25 nm². Again, the differential conductivity was measured simultaneously with the topography. Figure 3.10b)-d) shows maps of the dI/dU-signal at three particular bias voltages: b) U = +0.4 V, c) U = -0.15 V, and d) U = -0.2 V. It is a striking fact that island B exhibits a dI/dU-signal which differs from the underlying island A best visible in Figs. 3.10c) and 3.10d). In total we found that approximately 5-10 % of all triangular islands grown at 300 K exhibit this property. In contrast, island C which is grown

²The bright spots reflect an increased LDOS at this particular sample bias which is induced by the local adsorption of hydrogen and which has already been described in Fig. 3.8b). However, we would like to mention that the total amount of hydrogen adsorbed on the surface is far less than 0.01 L (1 Langmuir = 1×10^{-6} torr·s).

on top of island B does not exhibit a significantly different dI/dU-signal compared to island A. One possible explanation for this change of the electronic structure with respect to the underlying (0001)-surface is a stacking fault between the surface and the first subsurface layer. It is also obvious from our spatially resolved data that the



Figure 3.10: a) Constant-current STM image of a sample surface similar to that shown in Fig. 3.2b) simultaneously exhibiting a large variety of local coverages. On top of the high Gd(0001) island (A) triangular shaped Gd islands (B and C) appear. Nucleation of second (D) and third (E) monolayer patches can be observed on the first monolayer of Gd/W(110). Maps of the differential conductivity dI/dU measured at different sample bias are shown in b) U = +0.4 V, c) U = -0.15 V, and d) U = -0.2 V. The sample temperature during the measurement was 117 K.



Figure 3.11: Tunneling spectra measured at T = 117 K for local coverages $\Theta_{\text{loc}} = 1, 2, 3, \text{ and } \ge 4$ ML ("island"). For local coverages of 2 and 3 ML the surface state exists but the exchange splitting is reduced. The spectrum of island B in Fig. 3.10a) is shifted by $\approx 20\text{--}30$ meV, possibly due to a stacking fault.

electronic structure of second and third monolayer patches differ from each other and that both differ from the Gd monolayer note to the sites marked by arrows D and E in Fig. 3.10a). The difference becomes clear in the tunneling spectra measured at sites with different local coverages as plotted in Fig. 3.11. To prevent an unwanted overlap the spectra have been shifted relative to each other. As already described for room temperature measurements [cf. Fig. 3.7b)] the Gd monolayer exhibits an asymmetric peak centered at a sample bias U = +0.3 V which does not represent a surface state. In contrast, the spectrum measured above a double layer patch shows two maxima of the dI/dU-signal: a distinct peak at $U \approx +0.45$ V and a weak shoulder at $U \approx -0.1$ V. Similar to the results previously shown in Fig. 3.8 we have performed experiments which show that both features vanish upon contamination (not shown here). Therefore we can conclude that the surface state already exists on Gd patches with a thickness of 2 atomic layers and down to an area of at least 25 nm^2 . If the local coverage is increased the occupied part of the surface state shifts to a higher binding energy. For $\Theta_{\text{loc}} = 3$ ML it amounts to -140 meV until it converges for $\Theta_{\text{loc}} \ge 4$ ML to the thick film limit of -180 meV. The binding energy of the empty part of the surface state exhibits a much weaker dependence on the local coverage. This behavior points to surface electronic properties being fully developed for films with a thickness of about four monolayers. The most likely explanation for this behavior is that the surface stress of epitaxially grown Gd(0001)/W(110) decreases with increasing coverage thereby approaching the equilibrium lattice constant of an unstrained hcp(0001)

surface. This explanation qualitatively agrees with an earlier LEED-study performed by Weller and Alvarado [19]. According to this reference the initial Gd surface stress $(8 \pm 3 \% \text{ at } \bar{d} = 3 \text{ Å})$ reduces by $\approx 75 \%$ by increasing the coverage to $\approx 4 \text{ ML}$ and is fully relaxed at $\approx 30 \text{ ML}$.

Temperature-dependent electronic properties of Gd(0001)

Recently the temperature-dependent electronic structure of Gd(0001) attracted considerable interest. Although it has been consistently described by experimentalists that the "exchange splitting" between majority and minority spin parts of both, occupied [28, 29] and empty [30] bulk states, vanishes at the bulk Curie-temperature $T_{\rm CB}$ the interpretation of these experimental data is a topic of intensive theoretical discussions. In spite of the fact that due to the local nature of the magnetic moments in RE metals an application of simple band-theories seems not to be justified this behavior of Gd(0001) has been called "Stoner-like" and described in the framework of mean-field theory by a decreasing influence of the local 4f moments on the conduction electrons [28,31]. On the other hand the terminology of an "exchange splitting" in the case of local magnetic moments was called "ill-defined" in a theoretical contribution by Sandratskii and Kübler [32]. According to this local-spin-density-functional approximation the local magnetic moment of the conduction electrons amounts to 50% of its value in the ground state even above the Curie-temperature which is not caused by a relative shift between both spin-components but rather due to unequal contributions of both spin parts to the spin-mixed states lying below the Fermi-level.

In the following we will focus on the temperature-dependence of the surface electronic structure of Gd(0001). Both, evidence for pure Stoner-like [27] and pure spinmixing behavior [29] was found. These experiments were performed by means of photoemission (PE) and inverse photoemission (IPE) spectroscopy which suffer from the limitation to electronic states below and above the Fermi-level, respectively, i.e. the exchange splitting³ $\Delta_{ex}^{\text{surf}}$ can only be determined by different experimental techniques not being carried out on the same sample. This experiment-inherent disadvantage carries additionally weight as this, first, multiplies potential systematic errors which may occur at the determination of the Fermi-edge in the spectra and, second, as the complex film morphology of Gd may cause different magnetic properties. Scanning tunneling spectroscopy (STS), in contrast, avoids both problems since it enables to detect the density of electronic states above sample surfaces below the tip apex on both sides of the Fermi-level in a single measurement and spatially resolved. Since the $5d_{z^2}$ -like surface state is located around the $\overline{\Gamma}$ -point of the surface Brillouin zone it exhibits a large fade-out length into the vacuum (cf. Ref. [20]) and is therefore ideally suited to be investigated by STS. Therefore, the temperature dependence of the exchange splitting of the Gd surface state can *directly* be determined. Additionally,

³Although it has been questioned by Sandratskii and Kübler [32] whether the energetical difference between the peak positions of the occupied and the unoccupied surface state can be considered as an exchange splitting Δ_{ex}^{surf} we will use this terminology in the following.



Figure 3.12: Typical tunneling spectra

measured above Gd(0001)-islands (black) and 1ML Gd/W(110) (grey) at 29 K, 171 K, 293 K and 357 K [topography similar to Fig. 3.2(b)]. Close to the ground state (T =29 K) the exchange splitting between the occupied and unoccupied surface state above Gd(0001)islands is \approx 700 meV. While the binding energy of the electronic feature measured above the Gd monolayer remains unchanged, the exchange splitting of the surface states decreases with increasing temperature.

the spatial resolution enables to evaluate the influence of morphology and local film thickness.

As mentioned above it is still under debate whether or not the binding energy of both parts of the surface state changes with temperature. To unravel this open question we have performed scanning tunneling spectroscopy at variable temperatures. Figure 3.12 shows typical tunneling spectra as measured above the monolayer (grey curve) and three-dimensional islands (black curve) of a Gd-film grown in the Stranski-Krastanov-mode at four selected temperatures: close to the ground state (T = 29 K), slightly cooled (T = 171 K), at room temperature (T = 293 K), and at elevated temperature (T = 357 K). Close to the ground state (T = 29 K), both peaks being characteristic for the occupied and empty part of the surface state can clearly be observed at U = 460 mV and U = -240 mV, respectively. As the temperature is increased from T = 29 K up to 293 K both peaks obviously shift towards the Fermi-level, i.e. zero bias. This observation is in strong disagreement with a pure spin-mixing behavior as proposed on the basis of PE-experiments performed by Li *et al.* [29]. However, increasing the temperature above 293 K does not lead to a further shift of both peaks. Unfortunately, the binding energy of the occupied surface state could not be determined above T = 360 K. This is caused by the background of the differential conductivity which rises to negative sample bias as indicated by a hatched line in Fig. 3.12. As the temperature is increased this background extends its tail above the Fermi-level. Thereby it becomes more and more difficult to determine the peak position of the occupied surface state and the shoulder vanishes above 360 K. For comparison we have also plotted the monolayer spectra. The asymmetric peak does not shift significantly in the temperature range under study.

Figure 3.13 summarizes our temperature-dependent results on the binding energy of the occupied and the unoccupied part of the surface state, respectively, as measured above Gd-films of different local coverage Θ : Gd-films with $\Theta > 30$ ML (*), Gdislands with 4 ML $< \Theta < 15$ ML (\Box), bilayer islands (\circ), and the Gd-monolayer (\diamond). In the temperature range between 20 K and 380 K the binding energy of the unoccupied (occupied) part of the surface state as measured above Gd-islands (\Box) shifts from 490 meV (-220 meV) towards 300 meV (-60 meV). In contrast, the peak position of the unoccupied electronic state above the first monolayer remains unchanged (cf. Fig. 3.12). Our results are in good agreement with an earlier temperature-dependent PE and IPE study [27] up to 300 K. Within our energetical resolution ($\pm 50 \text{ meV}$) the high temperature STS-data also fit PE measurements from the same reference. However, a significant discrepancy exists for IPE results obtained above 300 K [27, 30]. Our STS-data reveal that the binding energy of the unoccupied surface state is 300 meV instead of 20 meV above 350 K [27]. The results indicate no difference between dI/dU-spectra of smooth Gd-films (\star) and Gd-islands (\Box). Furthermore, samples as shown in Fig. 3.2c) enabled to measure also the peak positions above Gd bilayer islands on $W(110)(\circ)$. While the binding energy of the unoccupied surface state is basically the same for bilayer islands and three-dimensional Gd-islands, the binding energy of the occupied surface state is diminished by approximately 100 meV at temperatures below 250 K (cf. Fig. 3.11) and by approximately 50 meV above room temperature.

The temperature-dependence of the exchange splitting $\Delta_{\text{ex}}^{\text{surf}}$ between the peak positions of the occupied and the unoccupied surface state, as measured above various Gd-films (cf. Fig. 3.2a)-c)), is plotted in Fig. 3.14. For thick Gd-films we measured a splitting of about 700 meV close to the ground state. The splitting reduces with increasing temperature in the temperature range between 20 K and 300 K but does *not* approach zero up to the maximal temperature in this study, i.e. 360 K. The exchange-splitting measured above the two atomic layers high Gd islands amounts to only 600 meV at 20 K and also reduces continuously down to approximately 370 meV at 300 K. While the transition from decreasing to constant exchange-splitting might be interpreted in terms of a Curie-temperature T_{C} this is strictly excluded for a film thickness as low as 2 ML since at this coverage T_{C} is far below 100 K [33]. Obviously, our experimental data can neither be described within a pure Stonernor within the pure spin-mixing-picture. However, one should keep in mind that the magnetism of Gd is caused by localized magnetic moments resulting from the halffilled 4f-shells which are coupled via RKKY-interaction mediated by spin-polarized 5d conduction electrons. At least qualitatively the differences in the temperature-



Figure 3.13: Plot of the binding energy versus sample temperature for surface states as measured above (\Box) thick Gd islands ($\Theta \geq 5$ ML), (\star) Gd films ($\Theta \geq 30$ ML), (\circ) Gd bilayers, and (\diamond) the Gd monolayer. Every data point represents an average of several measurements. The overall error bars are estimated to ± 50 meV. The binding energy of the electronic feature above the Gd monolayer remains constant within the experimental error. The exchange splitting between occupied and unoccupied surface states is smaller for the bilayer than for thicker films. This is caused mainly by a reduced binding energy of the occupied surface state.



Figure 3.14: Temperature–dependent exchange splitting of the hcp Gd surface state.

dependent behavior of the exchange-splitting between 5d bulk and surface states can be explained by their different degrees of localization [29]. While the bulk 5d conduction electrons are delocalized the $5d_{z^2}$ -like surface state is highly localized even within the surface plane [20, 22]. As already described by Sandratskii and Kübler [32] the magnitude of the atomic 4f magnetic moment is constant with temperature. In the ground state Gd is ferromagnetic and all 4f-moments are parallely aligned. Rising the temperature leads to fluctuations of the relative axes of adjacent moments which have been theoretically modelled in the past by spin-waves with the wave vector \vec{q} , i.e. the atomic moment of the $(n + 1)^{\text{th}}$ atom is rotated by a certain angle α with respect to the n^{th} moment (Fig. 3.15). In this situation the degree of localization is of importance: the less its localization the more the 5d-electron experiences the collec-



Figure 3.15: The 4f-shell of each Gd atom carries a local magnetic moment which remains constant with temperature. The relative axes of these magnetic moments fluctuate with rising temperature. This behavior can be modelled by spin-waves being characterized by the angle α between nearest neighbors.

tive 4f polarization which decreases the resulting exchange-splitting until it vanishes when the so-called magnetic correlation length exceeds the spiral periodicity $|\vec{q}|^{-1}$ Our experiments show that the exchange–splitting of the Gd(0001) surface state reduces with increasing temperature between 20 K and approximately 300 K but does not vanish up to a temperature of 360 K being well above any possible surface Curietemperature. In the framework of the spin-wave model already mentioned above and considering only nearest neighbor interactions—which is sufficient in most cases—this observation suggests that the 4f-moments of adjacent Gd atoms turn their relative quantization axis from parallel ($\alpha = 0$) to an unordered configuration ($\langle \alpha \rangle = \pi/2$) between 20 K and 300 K thereby going from minimal to maximal disorder. Further increasing the temperature has no effect on the spin-order between nearest neighbors and consequently the exchange-splitting remains constant since the high spatial localization of the surface state helps to maintain a local exchange splitting due to interaction with the atomic 4f magnetic moments at temperatures which exceed the Curie-temperature. The same argument can also be applied for the results obtained on Gd films as thin as 2 ML.

Spin-polarized tunneling

Figure 3.16 schematically illustrates the principle of spin-polarized scanning tunneling spectroscopy (SP-STS). A sample which exhibits an exchange-split surface state with a relatively small exchange splitting is ideally suited for our experimental approach. If the exchange splitting Δ_{ex} is too large one spin-component would be too far from the Fermi-level and not accessible by STS, as e.g. in the case of Fe(001), where Δ_{ex} amounts to 2.1 eV and only the minority band appears as a peak in the dI/dUspectra just above the Fermi-level [34]. In contrast, the majority (minority) part of the Gd(0001) surface state at 20 K has a binding energy of -220 meV (500 meV), i.e. the exchange splitting only amounts to 700 meV far below the Curie-temperature of 293 K.

In the following we consider vacuum tunneling between a Gd(0001) surface and a tip material for which the sign of the spin-polarization does not reverse in the energy range of interest, i.e. ± 0.5 eV around the Fermi-level. This condition is fulfilled for Fe [35]. For simplicity a constant spin-polarization is assumed [Fig. 3.16(a) bottom]. If the magnetization direction of the tip remains constant we have two possible magnetic orientational relationships between tip and sample, parallel or antiparallel. Since, however, both the majority and the minority component of the Gd(0001) surface state appear in our tunneling spectra, in any case the spins of *one* component of the surface state will be *parallel* with the tip while the *other one* will be *antiparallel*. Therefore, the spin-valve effect will act differently on the two spin-components; due to the strong spin-dependence of the density of states the spin-component of the surface state parallel to the tip magnetization is enhanced at the expense of its counterpart being antiparallel. Consequently, by comparing tunneling dI/dU-spectra measured above domains with opposite magnetization we expect a reversal in the contrast at


Figure 3.16: (a) The principle of SP-STS using a sample with an exchange split surface state, e.g. Gd(0001), and a magnetic Fe tip with a constant spin polarization close to E_F : due to the spinvalve effect the tunneling current of the surface state spincomponent being parallel to the tip is enhanced at the expense of its spin-counterpart. (b) This should lead to a reversal in the dI/dU-signal at the surface state peak position upon switching the sample magnetically. (c) Exactly this behavior could be observed in the tunneling spectra measured with the tip positioned above an isolated Gd-island (see arrow in the inset).

the majority and minority peak position [Fig. 3.16(b)].

Tunneling spectra measured in an external magnetic field with an Fe-coated probe tip ($\Theta_{tip} = 10 \text{ ML}$) positioned above an isolated Gd(0001) island show exactly the expected behavior [Fig. 3.16(c)]. After inserting the sample in the STM sample holder and cooling down to 70 K it was magnetized in a magnetic field of +4.3 mT applied parallel to the sample surface. Subsequently, 128 tunneling dI/dU-spectra were measured in remanence with the tip positioned above the Gd-island marked by an arrow in the inset of Fig. 3.16(c). Then the direction of the magnetic field was reversed (-4.3 mT) and further 128 tunneling dI/dU-spectra were measured at the same location. This procedure was repeated several times. Figure 3.16(c) shows the averaged tunneling spectra measured in remanence after the application of a positive or negative field. Comparison of the spectra reveals that for positive field the differential conductivity dI/dU measured at a sample bias which corresponds to the binding energy of the occupied (majority) part of the surface state is higher than for negative field. The opposite is true for the empty (minority) part. We have chosen free-standing Gd islands on W(110) for this experiment since it is known from Kerr-effect measurements [36] that the coercivity is only 1–1.5 mT, i.e. much lower than the applied field. Therefore, one can safely conclude that the magnetization of the sample was switched by the external field while the tip magnetization remained unchanged.

The high spatial resolution down to the atomic scale is the special merit of scanning tunneling microscopy and spectroscopy. We performed spatially resolved measurements at T = 70 K with a W-tip coated with 5–10 ML Fe on a sample prepared by depositing 10 ML of Gd on the W(110) substrate held at 530 K. This preparation procedure leads to partially coalesced Gd-islands with a Gd wetting layer on the W(110) substrate.

The images in Fig. 3.17 were obtained on a sample in the magnetic virgin state,



Figure 3.17: dI/dU-image at the majority [(a) U =-0.2 V] and the minority [(b) U = +0.45 V] surface state peak position. While a strong contrast within the Gd-island can be recognized in (a), a weaker and opposite contrast is present in (b). (c) In the asymmetry image the contrast is enhanced and tip changes less visible.

i.e. it was not magnetized by an external field. The scan range is $2\mu m \times 2\mu m$ with 250×250 pixel² resolution. At every pixel a dI/dU-spectrum was recorded. In Fig. 3.17 dI/dU-images are shown measured at (a) U = -0.2 V and (b) U = +0.45 V, i.e. sample biases which correspond to filled and empty parts of the surface state, respectively. Since the surface state does not exist on the heavily strained first monolayer of Gd/W(110) its differential conductivity is much lower than above fully relaxed Gd(0001). Consequently, the first ML of Gd/W(110) appears black. Besides obvious tip instabilities (stripes along the fast scan direction) which are most apparent at negative sample bias and indicated by an arrow, a strong contrast on the Gd-islands in the dI/dU-signal at U = -0.2 V [Fig. 3.17(a)] and a weaker contrast of opposite sign at U = +0.45 V [Fig. 3.17(b)] is visible. In both cases the contrast originates from differences in tunneling spectra similar to those observed upon switching the sample magnetization by an external magnetic field [cf. Fig. 3.16(c)]. The image shows not always simple light and dark contrast but intermediate. This is possible if some of the magnetic domains have a magnetization which is not in line with the Fe spins. We therefore conclude that we observe the magnetic domain structure of the sample.

The influence of sporadic tip changes on the image can be reduced by introducing an asymmetry parameter A which is also used in other spin-sensitive techniques in order to eliminate instrumental artefacts. Let $I_{\text{maj,min}}$ be the intensity of the dI/dUsignal measured at the majority (minority) peak position, then

$$A \stackrel{\text{def}}{=} \frac{(I_{\text{maj}} - I_{\text{min}})}{(I_{\text{maj}} + I_{\text{min}})} . \tag{3.1}$$

Figure 3.17(c) shows the asymmetry image composed by calculating A at every pixel with Figures 3.17(a) and (b) as the input data. Obviously, this procedure reduces the influence of tip instabilities and enhances the domain contrast.

To evaluate the signal strength and the spatial resolution obtained so far we have zoomed into a detail, i.e. a particular Gd-island already visible in the top part of Fig. 3.17. Again the dI/dU-signal for both surface state spin-contributions is plotted in Fig. 3.18(a) (filled) and (b) (empty). Both images show a domain wall crossing the island from top to bottom. A closer inspection of line-sections drawn from left to right along boxes across this domain wall reveals that both the filled and the empty surface state spin-part increase and decrease in intensity on a lateral scale below 20 nm, respectively. The spin-polarization P of the majority (minority) part of the surface state can be estimated by

$$P = 1/S \cdot \frac{\left(I_{\text{maj},(\text{min})}^{\uparrow} - I_{\text{maj},(\text{min})}^{\downarrow}\right)}{\left(I_{\text{maj},(\text{min})}^{\uparrow} + I_{\text{maj},(\text{min})}^{\downarrow}\right)}, \qquad (3.2)$$

where $I_{\text{maj},(\text{min})}^{\uparrow}$ and $I_{\text{maj},(\text{min})}^{\downarrow}$ are the intensity of the dI/dU-signal measured at the majority (minority) peak position above different magnetic domains (\uparrow,\downarrow) and S is the spin polarization of the tip. Assuming S = 0.44 [12] we obtain up to $P \approx 0.45(-0.24)$



Figure 3.18: Detail from the data set shown in Fig. 3.17. (a) In the dI/dU-image measured at U = -0.2 V the left part of the island is brighter than the right part indicating a parallel alignment of the filled surface state spin-part with the tip spin-polarization at the Fermilevel. (b) dI/dU-image measured at U = +0.45 V. (c) Plot of the dI/dU-signal drawn along the line-sections indicated above. (d) Asymmetry image calculated from (a) and (b).

for the majority (minority) part of the surface state, which is in good agreement with temperature dependent spin-resolved (inverse) PES data [9, 10]. We observed higher spin polarization values when using Fe coated tips which were magnetized by a permanent magnet instead of Helmholtz coils pointing to the fact that a magnetic field of $H \approx 8$ mT as produced by the latter is not sufficient to magnetize the tip to its saturation value.

The influence of the stray field of the probe tip on the domain structure of the sample is a critical parameter which has to be assumed in order to determine the significance of the experimental images. We never observed any indication of tip induced changes of the sample domain structure for Fe–coatings with $\Theta_{tip} \leq 10$ ML. If, however, $\Theta_{tip} \geq 100$ ML tip–induced magnetic modifications were observed. Figure 3.19 shows the asymmetry image of Gd(0001)–islands grown on a W(110) substrate.



Figure 3.19: Asymmetry image of a Gd sample measured with a 100–200 ML thick Fe– coating on the tip. The circles indicate two Gd–islands whose magnetization were accidentally switched probably due to interaction between tip and sample mediated via the stray field.

The sample was scanned from bottom to top with a tip coated with an Fe-film of 100-200 ML thickness. Two circles indicate locations where a checkerboard pattern was observed which is characteristic for a magnetization reversal of single Gd-islands. Magnetic switching of the tip can be ruled out since other Gd-islands are imaged unchanged in subsequent scan lines.

Summary

In summary, we have performed a detailed investigation of the surface electronic properties of Gd films on W(110) in dependence of film thickness, structure, temperature, and exposure to gaseous adsorbates. Based on this knowledge we could proof spinpolarized tunneling into the exchange split surface state of ferromagnetic Gd(0001). The spin-dependent nature of the signal could be proven by switching the observed contrast on Gd(0001) in an external magnetic field. We could image the domain structure of Gd(0001) islands with a resolution below 20 nm. The thickness of the magnetic tip coating plays an important role: changes of the sample domain structure were never observed for $\Theta_{tip} \leq 10$ ML Fe, but an Fe thickness exceeding 100 ML leads to magnetic switching of single Gd-islands on the sample.

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3.2.2 Magnetic Force Microscopy in Ultrahigh Vacuum Applied to Ultrathin Cobalt Films

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Introduction

In the past few years there has been growing interest in the magnetic anisotropy of ultrathin films. This attention originates both from fundamental physics and from applications in information storage technology. A strong effort is made to discover and develop new materials which could be used in future recording media. To observe the magnetic domain structure of as prepared thin films it is nessecary to perform the experiment under UHV conditions (cf. section 3.5.4), for the film must be prevented from oxidation and can not be covered by a protection layer since this might influence the magnetic domain structure.

The probe tip of the combined AFM/STM [1] we used and the sample could be exchanged under UHV conditions. Thus, we were able to test the Fe–coated tips used as probes in this study on known samples before measuring the Co thin films. Furthermore, we could measure the same Co film with several tips to confirm the results we have obtained. The maximum lateral scan range of the microscope is 6 μ m.

Cobalt on Silicon (001)

The first Co films were deposited on a Si (001) substrate. The Si was prepared by heating up to 700° C for several hours to remove the oxide layer. Subsequently a thin Co film was deposited on the silicon substrate under UHV conditions. Figure 3.20 shows two noncontact images of a 15 nm thick Co film. It reveals a polycrystalline structure with crystallites of a diameter of 7 nm and a rms roughness of 0.45 nm. Since the size of the crystallites is of the same order as the imaging tip the measured diameter gives the upper limit of the true size and the rms roughness is a lower limit, respectively.

We applied MFM to study the domain structure using silicon cantilevers with integrated tips coated with 20 ML of iron (cf. section 3.5.4). Figure 3.21 shows an example of an MFM image of an as-prepared Co film of 15 nm thickness. The data were obtained in *plane subtraction*-mode (PS-mode) measuring force gradients at a tip-sample separation of 35 nm. Structures with two bright and two dark parts joined on one edge, as marked by the circle in figure 3.21, occur several times. They are part of a domain wall of a complicated in-plane domain structure. According to a model of Huber and Middlehoek [2,3] a 180° domain wall of an in-plane magnetised thin film is replaced by a system of 90° Néel walls and Bloch wall segments (Bloch lines). At every second Bloch line the direction of the magnetisation changes more rapidly forming a



Figure 3.20: Noncontact AFM images of a 15 nm Co/Si thin film reveals a polycrystalline structure. The size of the crystallites is less or equal to 7 nm. The rms roughness was calculated to be 0.45 nm.



Figure 3.21: MFM images showing the cross-tie domain wall structure of a 15 nm Co/Si(100). For a detailed discussion see the text. Scan area of 4 μ m × 4 μ m (left) and 2.5 μ m × 2.5 μ m (right). Range of frequency shift of 3 Hz.



Figure 3.22: Simultaneously obtained topographic (left image, lower line scan) and magnetic data (right image, upper line scan). the scan area is 3.5 μ m × 3.5 μ m. Two squares were generated by measuring topography in AFM contact mode. These squares show up in topography as well as in the force gradient image as bright areas. The plotted lines show averaged data from the dashed boxes.

cross-tie (the marked structure). Since the cross-tie walls consist of a system of 90° Néel walls, the changes of the magnetisation direction induces magnetic charges which occur at different places [4,5]. As a result the cross-tie wall gives rise to a complicated stray field distribution above the surface of the sample. However, MFM is capable of resolving single cross-ties, due to its high sensitivity and lateral resolution [6].

Two phenomena were observed when a Co film was touched by the probe tip. The film expanded locally due to the stress applied by the tip. Simultaneously the area shows up bright in the force gradient data. The height of expansion depends on the applied force. For higher forces the film expanded more. The maximum expansion reached was about 2 nm. If the force was set too high the films were locally destroyed. The parameters required to expand Co films locally to a certain height, such as tip to sample contact time (speed of tip movement, respectively) and feedback setpoint, depended on the preparation of the Co film and the tip probe used. A certain lower force limit had to be exceeded to induce a change. Therefore it is possible to image the expanded structures even in contact AFM mode.

An example is shown in figure 3.22. Two squares were written by scanning two areas of 500 nm \times 500 nm in AFM contact mode. Using the *lift mode* one can exclude the change in force gradient signal to be due to the height of the topography. Two averaged line sections are displayed to the right in figure 3.22. The data were taken from the areas marked by dashed boxes on both images. Averaged data were shown for two reasons. First, the noise is reduced. Second, individual scan lines may differ in shape due to disturbances during the measurement. Since we are intrested in common features averaging is preferable. The line sections show a very sharp magnetic contrast whereas the topography is less distinct.

The lateral expansion of the bright area in the force gradient image also depended on the duration of the applied stress. It might well exceed the area of tip to sample contact. Figure 3.23 shows two examples for a Co film of 12 nm. The film was pressed by the tip almost in the centre of the image for several seconds, generating the bright



Figure 3.23: Touching effect on a 12 nm Co film. For the left image the tip was approached the sample surface in the centre of the scan area. After a few seconds it was moved to the lower left corner. The width of the written line (arrow) is 100 nm. For the right image the tip was moved in a more complex way. The data where measured in PS-mode.

spot. Then the tip was moved across the sample surface. The fine line, marked in the left image of figure 3.23, is the result of this movement. The movement was stopped three times for the right image. Finally the slope of the sample was determined and the image was scanned in PS-mode.

The observed structures in the force gradient images are due to the change of the magnetisation from in-plane to perpendicular direction. The nucleation of domains is due to an effect simular to magnetostriction. The magnetic moments are aligned in the direction of the force applied by the AFM tip when touching the surface. Although the forces in AFM are of the order of nN the pressure applied may well be in the range of GPa due to the small contact area of the probe tip providing enough energy to change the direction of the magnetisation. The size of the generated domains depends on the sample properties and on the way the sample was pressed by the tip. We found the contact time to be the key parameter for the domain extension.

To study the time dependence in more detail we touched a 60 nm Co film at several positions with increasing contact times. The results are displayed in figure 3.24. The times ranged from 1 second to 120 seconds. We were able to 'write' circular domains. The diameter of the nucleated domains increases with the contact time. These findings were confirmed on samples of 17 nm and 23 nm of Co. For similar contact times and similar forces of about 60 nN we got the same size of the written domains. The induced domains were found to be stable for at least 3 days.



Figure 3.24: MFM images of 'written' domains using different time periods for the tipsample contact. The time periods were for the left image of 1, 2, 4 (lowest row from left to right), 8, 12, 16 (middle row from right), 20, 32 and 64 (upper row from left) seconds. For the image to the right 80, 100 and 120 seconds were used (bottom to top). The diameter of the domains increases from about 64 nm to 290 nm in the left image and from 350 nm to 440 nm in the right image.

Summary

Thin Co films on Si (100) substrates have an in-plane domain structure with crosstie domain walls. The film expands when it is pressed by the probe tip. Then a magnetic domain is formed with magnetisation perpendicular to the sample surface. The domain grows as a function of the contact time.

Cobalt on Gold (111)

Ultrathin Co films on Au (111) substrates reveal a magnetic reorientation transition from out-of-plane to in-plane anisotropy at a thickness between 3 and 5 monolayer (ML). This interesting property has been previously observed by scanning electron microscopy with polarization analysis (SEMPA) [7,8]. Since the perpendicular anisotropy in magnetic thin films finds an application in perpendicular recording the observation is of great interest to be explored in more detail.

In the present study we have applied magnetic force microscopy (MFM) under ultrahigh vacuum (UHV) conditions in order to examine the magnetic domain structure of ultrathin Co/Au (111) at high spatial resolution as a function of film thickness and carbon content. While our MFM observations of the magnetic reorientation transition for clean Co films are in full agreement with earlier SEMPA studies we have found that the presence of carbon can significantly defer the reorientation transition by at least a factor of four. This surprising result may have applications using ultrathin carbon–containing Co films for high–resolution perpendicular magnetic recording.



Figure 3.25: a: STM image of the clean Au (111) surface exhibiting the $(23 \times \sqrt{3})$ herringbone reconstruction. b: early stage of Co film growth, as used to estimate the evaporation rate of Co. The Co islands of a height of two monolayers decorate the elbow sites of the underlying Au (111) herringbone reconstruction. Scan size: 200 nm \times 200 nm.

An Au (111) single crystal was used as the substrate. The miscut was specified to be better than 2°. Hence, the average size of the atomically flat terraces is at least 6 nm. The crystal was cleaned by cycles of Ar⁺-ion-etching at an ion energy of 600 eV and subsequent heating to 450 °C. The surface was checked by Auger, LEED and STM. As soon as LEED indicated the $(23 \times \sqrt{3})$ reconstruction of the Au (111) surface [9] and STM revealed a clean surface (Fig. 3.25 a), the further preparation was stopped.

We deposited the Co on Au (111), as well as the Fe on the Si cantilevers, by electron beam evaporation. The flux rate of the Co or Fe, which is a measure for the evaporation rate, could be kept constant by automatically controlling the high voltage on the target. Hence, we determined the film thickness from the flux rate and duration of evaporation. For calibrating the Co evaporator, we prepared a sub-monolayer film on the Au (111) single crystal and measured the actual coverage by STM to get the evaporation rate (Fig 3.25 b). Co starts growing in islands of 2 ML at the elbows of the $(23 \times \sqrt{3})$ herringbone reconstruction [10]. The rate was found to be quite stable, if the filament current as well as the current between the filament and the evaporant was kept constant. The typical evaporation rate was about 0.6 ML/min.

The magnetic anisotropy of ultrathin Co films on Au (111) changes with increasing film thickness in the regime of a few ML. We measured the thickness dependence of the magnetic reorientation using two ways of sample preparation. One way was to prepare the Co films starting from a clean Au substrate each time. The second method was to increase the film thickness in steps by evaporating Co on top of an existing film.

In the first case our results met the observation of the other authors, who found the transition to take place at about 4 ML of Co. Our results are displayed in Fig. 3.26. The images were obtained with several different cantilevers. The first five images show



Figure 3.26: MFM images obtained at increasing Co film thicknesses. The numbers denote the actual film thickness in monolayers. Scan size: $5 \ \mu m \times 5 \ \mu m$.

MFM data for a Co film thickness of 1.8, 2.0, 3.0, 3.2 and 3.8 ML. For 1.8 ML only a weak contrast is visible, which will be discussed later. The next four images show areas with dark and bright contrast. This feature can be interpreted as the magnetic domain structure of a film magnetized perpendicular to the surface. The size of domains first increases (film thickness from 2 to 3 ML), then decreases as approaching the magnetic reorientation of the anisotropy, as shown for 3.2 and 3.8 ML.

For the next three images, which represent a Co film thickness of 4.1, 4.3 and 4.5 ML respectively, the appearance changes. Now, lines of dark and bright contrast can be seen. The contrast becomes more clear with increasing film thickness of 4.6, 4.8, 6.0 and 7.0 ML in Fig. 3.26. Since the tips were magnetized perpendicular to the sample surface, such a contrast indicates an in-plane magnetic domain structure. The image of the 6.0 ML film of cobalt shows an additional contrast within the magnetic domains which can be interpreted as a magnetic ripple structure [2, 11].

It is an interesting question for ultrathin Co films on Au (111) at which coverage the ferromagnetic order occurs and what type of domain structure can be observed. The first image in Fig. 3.26 shows MFM data at the minimal thickness of the Co film, where we obtained a magnetic signal in addition to topographic crosstalk. We

Figure 3.27: STM and MFM images of a Co film of 1.8 ML thickness. The images a) and b) show STM data of 400 nm × 400 nm and 100 nm × 100 nm scan size. Images c) and d) display MFM data of the same sample of 3 μ m × 3 μ m and 1.5 μ m × 1.5 μ m scan size. For comparison, image a) is displayed as an inset in image d) at the same scale.

examined this sample in more detail. Figures 3.27 a)-b) show STM data of a scan range of 400 nm × 400 nm and 100 nm × 100 nm. The Co film consists of islands of 2 ML height, where the third and fourth layer already started to grow. The islands began to coalesce, but there are still single islands as well as fractions of uncovered Au. Two MFM images of 3 μ m × 3 μ m and 1.5 μ m × 1.5 μ m scan area are displayed in Fig. 3.27 c) and d). In the upper right corner of image 3.27 d), image a) is displayed at the same scale as image d) for comparing the sizes of the magnetic and topographic features. The MFM images reveal bright and dark stripes of about 50–100 nm width, running mainly from the bottom left to the upper right corner. The stripes follow basically the steps of the Au (111) crystal.

To confirm the magnetic origin of these stripes and exclude possible electrostatic effects, the same sample was measured using a non-magnetic tip. We had to measure as close to the surface as possible to get any signal at all. The only structure found revealed features of significant smaller size than the magnetic structures and showed a stronger distance dependence. This structure is likely to originate from *van der Waals* interaction between the tip and the sample surface. The difference to the structure as

Figure 3.28: Intensity of the C(272) and Au(165) Auger line with respect to the intensity of the Co(656) Auger line plotted against time for a 5 ML Co film.

measured with the magnetic tip on the same sample affirm the magnetic origin of the contrast in Fig. 3.27 c) and d).

The observation of the same Co film for a longer period of time revealed a change of its magnetic structure. For Co films at a thickness just above 4 ML we achieved a reversal of the magnetic anisotropy from in-plane to out-of-plane direction. Former work indicates two possible reasons for this change. Speckmann et al. showed for a Co wedge on Au (111) that heating the sample leads to a shift of the critical thickness for the reorientation transition. This was due to the migration of Au to the sample surface [8]. Hope et al. found for a Co film on Cu, that the easy axis of magnetization could be switched by 90° within the film plane by CO dosing [12].

To find out the reason for the change of the magnetic structure in time in our case, we measured MFM and Auger spectroscopy on a 5 ML thick Co film for several days. From the Auger data the peak heights of two Auger lines, Au(165) (h_{Au}) and Co(656) (h_{Co}) , were extracted. The quotient h_{Au}/h_{Co} is plotted versus time in Fig. 3.28. Each mark represents the averaged data of five Auger spectra taken within one hour of time at different spots on the sample. The error bars denote the statistical error. The straight lines show extrapolated data as a guide to the eye. For the migration of Au to the surface the h_{Au}/h_{Co} -signal should rise in time. This can clearly be excluded by the data. The second data set in figure 3.28 shows the relation of the C(272) Auger peak height to the Co(656) peak height (h_C/h_{Co}) . Here we see that the Co surface becomes contaminated by C.

Along with the Auger spectroscopy MFM measurements were performed to observe the domain structure and thus to deduce the magnetic anisotropy. The letters in the lower part of Fig. 3.28 correspond to the MFM images in Fig. 3.29. The scan areas of

Figure 3.29: Evolution of the magnetic structure of a 5 ML Co film in time due to C contamination. The images were taken about 2, 6, 7, 8, 8.2 and 13 days after the sample preparation. Scan size: 5 μ m × 5 μ m.

the images are 5 μ m × 5 μ m each. The images a) and b) show domain walls indicating that the film is magnetized in plane. For image c) small perpendicular domains start to be seen, which became more pronounced in figure d). The sample ends up with a perpendicular domain structure as shown in figure e) and f). Comparing the MFM data to the plot in Fig. 3.28 one observes that the change in the MFM images corresponds to the rise of the C signal.

We studied the reversal effect in more detail by measuring the same area of 5 μ m × 5 μ m on a 4.1 ML Co film sample several times. The CO concentration within the UHV chamber was increased during the experiment. The results of the MFM measurements are displayed in Fig. 3.30. The images a)–e) were taken 50 min, 90 min, 130 min, 170 min and 210 min respectively after the sample preparation. Image f) was taken the next day, 630 min after preparation. The first image shows two domain walls running parallel from the bottom left to top right corner indicated by arrows. A third wall is visible in the lower right corner. The second image shows the same three domain walls. For the images from c) to e) bright and dark areas arise, indicating the nucleation and growth of domains magnetized perpendicular to the sample surface. Finally, a perpendicular domain structure is observed the next day at the same area, as shown in image f). This time the magnetization reversal went much faster which is due to the increase of the CO concentration.

As mentioned before, we used a second way of sample preparation to study thickness dependent effects. Starting from a 2 ML film we deposited Co on top, but without cleaning the Au (111) crystal between the evaporation steps. We found a change of the magnetic structure, but at a significantly higher thickness of the Co film. The series of MFM-images (Figure 3.31, 5 μ m \times 5 μ m each) starts at a Co film thickness of 13 ML. For each image an additional monolayer of Co was deposited. The first six images show clearly a perpendicular magnetization. The average width of the magnetic domains changes drastically from about 600 nm down to 100 nm. From 19 ML on no clear perpendicular domain structure is visible. For thicker Co films we get some indication of a magnetic ripple structure (Fig. 3.31, 26 - 28 ML), but no clear indication of a domain wall is observed. The samples with a Co film thickness of 16 -21 ML were also measured at a scan size of 2.5 μ m \times 2.5 μ m. The result is displayed in figure 3.32. Up to 18 ML a domain like pattern can be seen. The next image shows no structure from perpendicular domains, but some influence from the tip on the magnetic structure is present indicating that the Co film becomes magnetically soft at the critical thickness for the reorientation transition. After reaching a Co film thickness of 28 ML we performed Auger measurements. We see a huge C(272)-peak, but still signatures of Co and Au. From the series of MFM data presented in Figs. 3.31

Figure 3.30: Evolution of the magnetic structure of a 4.1 ML Co film in time due to C contamination observed at the same spot on the sample, 50', 90', 130', 170', 210' and 630' after sample preparation. Scan size: $5 \ \mu m \times 5 \ \mu m$.

and 3.32 we conclude, that the carbon contaminated Co film exhibits perpendicular anisotropy up to a thickness of at least 18 ML.

Conclusion

We observed the reorientation transition of the magnetic structure of ultrathin Co films on Au (111) by means of MFM as a function of the Co film thickness. At a film thickness of 1.8 ML we were able to detect a magnetic contrast of striped domains running parallel to steps on the Au substrate. The width of these stripes was about

Figure 3.31: MFM images of a sample for which the thickness of the Co film was increased in steps of 1 ML starting at 13 ML. Again, the numbers denote the Co film thickness in monolayers. The magnetization is perpendicular up to 18 ML due to C contamination. Scan size: 5 μ m × 5 μ m.

Figure 3.32: Close up views of the deferred magnetic transition (cf. Fig. 3.31) for the 16 ML to 21 ML Co films contaminated by C. Scan size: 2.5 μ m × 2.5 μ m.

50-100 nm. Increasing the film thickness leads to a growth of the average domain size up to several μ m at 3 ML of Co. The domain size shrinks when approaching the critical thickness of 4 ML. For thicker films we observed a domain wall contrast indicating a magnetization in the film plane.

Carbon was found to have a huge influence on the magnetic domain structure of cobalt thin films. Due to the C contamination, the easy axis of the magnetization of in-plane magnetized Co/Au (111) films is switched back to perpendicular direction. Furthermore, the presence of C stabilizes the perpendicular anisotropy up to a Co film thickness of at least 18 ML.

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3.2.3 Micromagnetic Properties and Magnetization Switching of Single Domain Co Dots

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Introduction

Nanostructured ferromagnetic particle arrays have become of increased interest because of their possible application as high-density magnetic storage media [1, 2] and their potential as semiconductor-ferromagnet hybrid systems [3, 4]. The storage density of thin magnetic film recording media is among other reasons limited by domain walls and ripple structures which contribute to noise in the readback signal [5]. By using single-domain ferromagnetic particles as storage media this limitation can be overcome. Theory suggests that ferromagnetic particles below the critical size are in a single domain state [6, 7].Magnetization of a single-domain particle can be in two discrete states. Therefore each particle is able to code a single bit in a microstructured recording media [1].

To study the average magnetic switching behaviour of such single-domain particles, hysteresis curves of large arrays of particles have been taken using magnetometer [2,8]. Magnetic Force Microscopy [9] provides an excellent opportunity to image the magnetization of individual dots because of its high resolution and sensitivity [10,11].

We report on an investigation of the micromagnetic properties of Co-dot arrays on GaAs substrates in presence of an external magnetic field. The experimental results are complemented by micromagnetic simulations.

Experimental

Large arrays of ferromagnetic Co dots were fabricated on GaAs substrates using electron beam lithography [3,4]. In order to improve adhesion, the GaAs-substrates were covered with a Cr-layer of 3 nm thickness.

The structures were written by an electron beam into two layers of photo-resist material. After developing the exposed PMMA resist Cobalt was evaporated. After lift-off in acetone dot arrays were obtained. The dots have diameters in the range of 100 nm to 300 nm and a thickness of 4 nm to 17 nm. They were studied with MFM using a Nanoscope IIIa system [12] in the dynamic 'Lift mode'. We used Si cantilevers with integrated tips and spring constants between 3 N/m and 5 N/m. The resonance frequencies were found to be 55 kHz to 73 kHz. A single layer of 28 nm iron served as a magnetic thin film coating for the tips. Magnetic force gradients were measured by detecting phase shifts in the cantilever oscillation due to attractive or repulsive forces acting on the ferromagnetic MFM tip. We added an electromagnet for applying fields

in plane of the sample. The microscope was placed in the center of a Helmholtz coil pair. In this manner we were able to apply fields up to 60 kA/m.

In order to study the switching behavior we used the following procedure. The scan area is chosen so that only the dot we want to switch the magnetization is scanned. We use the same parameters as in all other measurements presented here. After an image of the initial state is obtained the external field is switched on while the scanning is continued. In this way the dot is influenced by the tip's stray field and the external field at the same time. The external field is increased until the magnetization direction of the dot changes. Then the external field is turned off. The dot is scanned again to see if the magnetization remains stable without the external field.

Results and Discussion

Figure 3.33 shows the topography of $3\mu m \times 3\mu m$ area and the corresponding MFM image of a Co dot array with a height of 7 nm. The dot diameter is 180 nm as determined from topographic measurements. The MFM image is characterized by a bright and dark contrast over each Co dot. The MFM tip was magnetized in z-direction (perpendicular to the plane of the sample) prior to the measurement. Therefore the magnetic contrast can be explained by an interaction between the in-plane magnetized dots and the MFM tip. For a Co dot height of 7 nm in an as-prepared state with a circular shape we did not observe any preferential alignment of the dots' magnetization direction.

Changing the geometry of the dots led to a change of the magnetization direction. For elliptically shaped Co dots the magnetization direction is aligned parallel to the long axis even for an as-prepared sample. Figure 3.34 presents the topography and the corresponding MFM image of an as-prepared sample with elliptical shape of Co dots. The lateral dimensions of these dots are 140 $nm \times 250$ nm and the height is 7 nm. The magnetization directions of the dots are aligned parallel to the long axis

Figure 3.33: a) Topographical and b) MFM-image of circular Cobalt dots (diameter 150 nm, thickness 7 nm) in as prepared state. The magnetic dipole contrast indicates that the dots are magnetized in plane. Scan-size: $2\mu m \times 2\mu m$

Figure 3.34: a) Topographical and b) MFM-image of elliptical Cobalt dots (140 nm \times 250 nm, 7 nm thickness) in the as prepared state. The magnetization of most dots is parallel to their long axes. Scan-size: $6\mu m \times 6\mu m$

(x-direction). Only a few dots revealed a different orientation of the magnetization.

In order to explain the magnetic contrast mechanism, we magnetized the Co dot sample in a strong external field (400 kA/m) outside the MFM system. The magnetic field was applied in-plane of the sample, parallel and antiparallel to the x-direction. Figure 3.35a shows a MFM image $(1.1\mu m \times 1.1\mu m)$ with Co dots in the remanent state. The dot array was magnetized by a field applied along the (-x)- direction (marked by a black arrow below image 3.35a) prior to the MFM-measurement. The magnetization directions of the dots are aligned in the direction of the applied field, bright contrast on the right side and black contrast on the left side are clearly visible. Applying the field in the opposite x-direction (marked by a black arrow below image 3.35b) led to an inverse magnetic contrast.

The MFM-measurements were performed in the dynamic mode with a phase detection system. Resonance frequency shifts towards higher frequencies lead to a bright contrast in the phase detection. Therefore, a repulsive interaction between tip and sample is represented by bright contrast and an attractive interaction by dark contrast. Repeating the same experiment with an in-plane magnetized MFM-tip led to a magnetic contrast which is characterized by an attractive interaction above the center of a dot and a repulsive interaction at both ends (Fig. 3.36 a). After the Co dot sample was magnetized in the opposite direction a repulsive interaction above the center of a dot and attractive interaction at both ends was observed (Fig. 3.36 b). In conclusion the remanent state of the Co dots could be described as a single domain. Applying an external field led to a preferential alignment of the magnetization direction of the Co dots. In order to study the magnetic switching behavior of individual dots we placed the microscope between a pair of Helmholtz coils. We could apply a field up to 60 kA/m in-plane of the sample.

The principle of this switching process is illustrated in figure 3.37. The applied field is chosen to be not that strong to change the magnetization of the dots. However the dots are also influenced by the stray-field of the tip. The magnetic coating of the

Figure 3.35: MFM-images of dots (diameter 200 nm, thickness 17 nm) that have been magnetized in a strong external field (400 kA/m) outside the microscope. After scanning a) the field has been applied in the opposite direction. The small figures show the results of a simulation for the MFM contrast of dots magnetized in (x)- and (-x)-direction (marked) and scanned by using a tip magnetized in z-direction (perpendicular to the plane of the sample). Scan-size: $1.1\mu m \times 1.1\mu m$

Figure 3.36: MFM-images of the dots used in figure 3.35 magnetized again in two opposite directions and scanned with a tip magnetized in *x*-direction. Attached are the results of the simulation for a tip magnetized in *x*-direction. Scan-size: $1.1\mu m \times 1.1\mu m$

tip is chosen in a way that its stray field does not switch the dot either. But if a dot is influenced by the superposition of the tip's stray-field and the applied field its magnetization direction should change.

In a first experiment [10] the magnetization of a single circular-shaped Co dot of 7 nm thickness has been switched. Figure 3.38 a) shows the dot in its initial magnetization state. Figure 3.38 b) was scanned with an external field of 100 Oe applied. The image shows that the magnetization direction of the dot has turned by 90° . The new magnetization state is stable even when the field is turned off as can be seen in Figure 3.38 c).

As the elliptical dots shown in figure 3.34 are preferably magnetized in two opposite directions, they are good candidates for switching experiments. Applying an external field parallel to the long axis of the sample with a strength of 400 kA/m led to a homogeneous alignment of all Co dots. For the local MFM switching experiment we chose an array of 3×3 dots. Figure 3.39 shows that the dots are all aligned in the initial state. Then the scan size is decreased so that only the dot in the middle is inside

Figure 3.37: Principle for magnetic switching of individual dots. The tip and the external field are chosen in a way that neither the stray field of the tip nor the external field are strong enough to cause a dot to switch. Only the superposition of the two leads to a change of the dot's magnetization.

the scanned area. While the dot is scanned the external field is increased until the magnetization direction of the dot changes. Another scan of all 9 dots taken without external field shows that the new magnetization state of the dot in the middle is stable and that the magnetization of the other dots has not changed. In the same way we switched the magnetization of each of the three dots on the right before scanning figure 3.39 c). Finally the dot on the left side as marked in figure 3.39 d) is switched. For each dot a different external field was needed to change its magnetization but every dot changed the magnetization direction by full 180° . For the five switched dots we used fields from 22 kA/m up to 40 kA/m. Other dots showed no change in their magnetization even at fields up to 60 kA/m.

The different switching behavior of the Co dots can be explained by their polycrystalline structure [11]. The net magnetization of a Co dot is given by the sum of

Figure 3.38: Series of MFM-images of a single circular Co dot. a)Initial magnetic state without magnetic field applied b) Image taken with an external field of 8 kA/m applied in the direction indicated by the arrow. The magnetization direction has turned by 90° c) Final magnetic state of the dot scanned without magnetic field applied.

Figure 3.39: Local manipulation experiment. After the dots were magnetized in a strong external field a) was scanned. Then the dot in the middle was scanned individually with a field of 20 kA/m applied. Figure b) shows that this dot has switched. The same way each of the three dots on the right has been scanned and switched individually with fields of 22 kA/m up to 40 kA/m. Finally the dot on the left (figure d)) was switched in the same way.

the magnetization of randomly orientated crystallites. The switching fields are also depending on the orientation and the magnetic coupling of the crystallites.

Micromagnetic Model

In order to investigate the switching mechanism of the Co dots in more detail a micromagnetic simulation [13] was performed.

The model consists of a tip which is positioned above a single elliptical Co dot. A rectangular coordinate system xyz is defined with the origin in the center of the dot and the x and y axis pointing along the long and short axis of the ellipse respectively. The tip is positioned at a height h_t above the dot and symmetrically with respect to the long axis of the dot. The tip is modeled as a nonmagnetic Si₃N₄ body coated with 28 nm Fe with a homogeneous magnetization M_t pointing in (-z)-direction. The tip's magnetization was chosen to be rigid as our experiments showed that the magnetization of the tips we used was not altered neither by the dot's stray-field nor by the

external field. The magnetic field of the tip was calculated analytically.

A 245 nm × 147 nm elliptical Co dot with a height of 7 nm was discretized into uniformly magnetized cubes with a sidelength of 7 nm. As a consequence of this discretization only an approximation to the elliptical dot shape is achieved. The magnetization of each cube m_i was assumed constant in magnitude ($M_d = 1.4 \times 10^6 A/m$) but allowed to rotate. All contributions to the interaction energy (the magnetic fields of the other cubes, exchange interaction, tip's stray field and external field) have been expressed as magnetic fields acting on the m_i . For the micromagnetic simulation an exchange energy constant of $A_d = 2 \times 10^{-11} J/m$ has been chosen [14]. Crystalline anisotropy was neglected to account for the polycrystalline nature of the dot.

In order to perform the simulation, all cubes of the dot were initially magnetized in the x-direction. Then the simulation was started. Each step of the simulation consisted of two parts. First the magnetic fields acting on each cube were computed. To save computing time the magnetic fields of the cubes were approximated by dipole fields for cubes separated by more than 21 nm and were neglected for cubes separated by more than 210 nm. Second, the Landau-Lifshitz-Gilbert equation was solved for each m_i using a fourth order Runge-Kutta method with adaptive time-steps. The simulation steps were carried out until a local energy minimum was reached. An indication for the minimum was taken that the maximum rotation of the m_i in a single simulation step was less than 0.1°. Additionally, the energy was computed separately and monitored.

The tip was positioned 100 nm right off the center on the long axis and an external field was applied. The simulation was carried out until the energy minimum was reached. The external field was then increased and the simulation of the dot was carried out again. This was repeated until the net magnetization of the dot was pointing in (-x)-direction. This external field was defined as 'switching field' H_{sw} and plotted as a function of the vertical tip-dot separation (Fig. 3.40).

Figure 3.40: This diagram shows the minimum field required to reverse the magnetization of a 147 nm \times 245 nm \times 7 nm Cobalt dot (switching field) as a function of the tip-dot separation.

Figure 3.41: Simulated stages of the magnetization configuration of the dot during the switching process. The arrows represent the x- and y-components of the dot magnetization at the position marked by the tail of an arrow. (a) the prepared single-domain state of the dot before any field is applied. (b)-(g) Intermediate stages of the switching process. The tip is positioned 50 nm above the surface of the dot and 100 nm toward its right end. An external field of 500 kA/m is applied in (-x)-direction. (h) Final stage, external field and tip are removed. The magnetization of the dot has switched.

The switching field increases with the tip-dot separation up to about 150 nm. If the tip is moved further away the switch is only due to the external field and not induced by the tip. After determining the switching field we visualized different stages of the switching process itself (Fig. 3.41).

The dot has been magnetized in a strong external field to align the magnetization of all cubes along the x-direction. Then the external field was turned off and the simulation was started to find the minimum of energy for the starting configuration of the switching process (Fig. 3.41 a)). Then the tip was positioned 50 nm above the dot and 100 nm right off its center. The previously calculated switching field has been applied. In figure 3.41 b) one can see a change of the magnetization along the (-x)-direction (upper-right and lower-right part of the dot). Two domain walls and a disturbance on the right can be seen. In figure 3.41 c) vortices show up which subsequentially move out of the dot (Figure 3.41 d)-g) while the area of homogenous magnetization along the (-x)-direction increases. In figure 3.41 g) the dot is mostly magnetized in the (-x)-direction except the right side where the influence of the tip's stray field is still visible. After the tip is removed and the external field is switched off the dot stays in a new single domain state (Figure 3.41 h).

Our micromagnetic simulations suggest that the switching process cannot be described by a uniform rotation of all magnetic moments in the dot (Stoner-Wohlfarth [15]) but by a nucleation and movement of vortices and domain walls caused by the superposition of the tip's stray field and the homogenous external field. This explains that the magnetization of a dot can be switched with lower external fields if the MFM-tip is scanning over the dot. A quantitative comparison shows that the switching fields determined by the simulation are about five times higher than those used in the experiment. Further investigation of the internal structure of the dots will probably be necessary to explain this difference.

Summary

In summary we have applied magnetic force microscopy to study the magnetic switching behavior of Cobalt dots. Dots of diameters below 250 nm and a height of 7 nm appeared to be in a single domain state with the magnetization lying in-plane. Elliptical dots are preferentially magnetized parallel to their long axis due to shape anisotropy. By combining an external field with the stray field of the ferromagnetic tip we changed the magnetization direction of a single dot by 180 °. From our micromagnetic simulations we conclude that the switching process is due to a nucleation of domain walls and vortices in the dot. Theoretical simulations do also show that the external field that is needed to switch the magnetization of a dot decreases when the magnetic tip is getting closer to this dot.

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3.3 Nanoelectronics

3.3.1 Low Temperature Scanning Tunneling Spectroscopy in High Magnetic Fields: The LDOS of N-InAs(110)

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Introduction

If one covers n-InAs(110) by several percent of an adsorbate, it exhibits a two dimensional electron gas (2DEG) at its surface [1–6]. The chemical nature of the adsorbate appears not to be important, but the exact position of the Fermi level with respect to the conduction band minimum depends critically on coverage and slightly on deposition temperature [1]. The adsorbate induced 2DEG at the surface combined with the low defect density on the InAs(110)-cleavage plane makes X/InAs(110) an ideal model system to study the interaction of a low dimensional semiconductor system with correlated electron materials like ferromagnets or superconductors [7]. The electronic interaction of the lateral interface can be studied by scanning tunneling spectroscopy (STS) at low temperature combining high lateral and spectral resolution. However, to understand STS data of X/InAs(110), it is first necessary to understand the clean system.

We investigated InAs(110) by STS at 8 K and in magnetic fields from 0 T to 6 T perpendicular to the surface. This section summarizes the main results. After describing the experimental setup and constant current images we deal with the problem of the tip induced band bending present in STS-measurements on semiconductor surfaces. Then we describe dI/dU-images obtained in zero magnetic field which show the scattering states of individual dopants located down to 20 nm below the surface. The analysis of the dI/dU-images allows to extract the energy dependence of the scattering coefficient of individual dopants. Then we describe dI/dU(U)-spectra, especially the part belonging to the density of states (DOS) of the conduction band (CB). By applying a magnetic field B > 2 T, we observe oscillations of the DOS of the CB in agreement with Landau quantization. Determination of the effective mass (m_{eff}) from the data reflects correctly the nonparabolicity of the InAs-CB. Laterally resolved images of the Landau levels show the dependence of its energetic positions on the local potential. A preliminary model is able to explain the corresponding energy shifts qualitatively. At the end of the article, we describe first results obtained on the system Fe/InAs(110). First we proof that Fe indeed pins the Fermi level in the conduction band, second we show that Fe is on the edge of mobility at room temperature and third we give some indications that Fe changes the electronic structure of the InAs(110)-surface in a non-simple way.

Experiment

The UHV-low-temperature STM apparatus is described in subsection 3.5.2 [8]. It allows measurements down to 6.2 K in magnetic fields up to 6.5 T perpendicular and 2 T parallel to the surface. The noise level in topographic images is below 5 pm and the spectral resolution is at least 7 meV. The thermal drift is about 3 Å/hour.

N-type InAs $(N_D = 2.0 \times 10^{16}/cm^{-3})$ is studied. The dopant density is checked by vander-Pauw-measurements at $T = 4 \ K - 300 \ K$ and Secondary Ion Mass Spectroscopy (SIMS). This reveals that the sample is degenerately doped and the main dopant is Sulphur. Hall and Shubnikov-de Haas-measurements proved that the sample exhibits no magnetic freeze out up to 8 T and down to 4.2 K [9].

STM and STS-measurements are performed after cleaving the sample in-situ at a base pressure below $1 \times 10^{-8} Pa$. Then the sample is transferred into the microscope, which is moved down into the cryostat within 5 min. The procedure results in a clean InAs(110)-surface with an STM-detectable adsorbate density of about $10^{-7}/\text{Å}^2$. Due to the low pressure inside the cryostat, it does not increase on the time scale of weeks. For the Fe/InAs(110)-measurements Fe is deposited by molecular beam epitaxy at room temperature with a rate of 1.2 ML/min. Again the sample is transfered into the microscope and moved down into the cryostat.

The STM-investigations are performed with an STM-tip (tungsten) priorly prepared on W(110) by applying voltage pulses (5 V/10 ms) between tip and sample. Topographic and dI/dU-images are recorded in constant current mode. The dI/dU-signal is recorded by lock-in-technique ($f = 1.5 \ khz$, $U_{mod} = 8.5 \ mV_{pp}$). dI/dU(U)-curves are measured at fixed tip position with respect to the surface. The tip-surface-distance is stabilized at a sample bias U_{stab} and a tunneling current I_{stab} as indicated. dI/dU is recorded by lock-in-technique ($f = 1.5 \ khz$, $U_{mod} = 3 \ mV_{pp}$). All measurements are performed at $8\pm 1 \ K$.

The Photoelectron Spectroscopy (UPS) measurements on Fe/InAs(110) are performed at the Hamburger Synchrotronstrahlungslabor (HASYLAB) with a VG ADES 400 spectrometer. Fe is deposited and the spectra are recorded at room temperature. Deposition rates are callibrated with a quartz balance.

Topographic Images

Fig.3.101a on page 141 shows an atomically resolved constant-current image of n-InAs(110). Atomic rows of one type of atom (presumably As) are visible. We did not observe the other type of atomic rows (In) in the voltage region U = -600 mV -400 mV. This is presumably due to the fact that the surface states of InAs(110) are located about 1 eV away from the band edges [10]. Fig. 3.42a and b show two larger scale images of the same surface. In Fig.3.42a one 0.5 Å high elevation with a full width at half maximum (FWHM) of 10 nm is visible in addition to the atomic rows. A number of such elevations of different heights and somewhat different FWHM is visible in Fig.3.42b. The elevations are caused by charged dopants [11, 12]. Indeed

Figure 3.42: a.) constant current image of n-InAs(110); 25 $nm \times 25 nm$, U = 150 mV, I = 75 pA; b.) constant current image of n-InAs(110) (3D): 500 $nm \times 500 nm$, U = 50 mV, I = 300 pA; c.) Simulation of a constant current image using the same parameters and magnifications as in b.) (see text for discussion).

we found that the number of elevations scales with the dopant concentration [13]. A simple model is able to reproduce the elevations qualitatively [14]. It assumes that the screened Coulomb potential of the ionized dopants leads to a local band bending, which shifts the DOS of the sample $\rho(E)$ according to:

$$\rho(E, x, y) \propto \sqrt{E - E_{CBM}(x, y)} \tag{3.3}$$

(E: energy; E_{CBM} : local energy of the conduction band minimum, x, y: surface position). The resulting apparent height in constant current images Δz is calculated by solving:

$$I \propto \int_0^{eU} \rho(E, x, y) \cdot \exp(-A \cdot \sqrt{(\hat{\Phi} - eU)/2} \cdot z(x, y)) \, dE \tag{3.4}$$

(e: electron charge, U: sample voltage). The effective barrier height $\overline{\Phi}$ can be obtained from I(z)-curves recorded at different sample voltages as described in eq. 3.6 or in [13]. Fig.3.42c shows a simulation obtained by distributing dopants according to the known dopant concentration randomly in a box $(600 \times 600 \times 50 \ nm^3)$. The potential is summed up at the surface to get $E_{CBM}(x, y)$ and respectively $\Delta z(x, y)$. Good qualitative agreement is achieved by the simple model. A more sophisticated model can be found in the literature [15]. As a rule of thumb, one should keep in mind that the apparent radius of dopants in STM-images as well as the depth up to which dopants are visible in STM-images is mainly the screening length λ in the corresponding semiconductor [16].

Analyzing dI/dU

At low temperatures $(T \approx 0K)$ and sufficiently large tip-surface distances, dI/dU(U,x,y) can be described by [17,18]:

$$\frac{dI}{dU}(U) \propto e\rho_{tip}(0) \cdot \rho_{sample}(eU) \cdot T(E = eU, U, z)
+ \int_{0}^{eU} \rho_{tip}(eU - E) \cdot \rho_{sample}(E) \cdot \frac{dT(E, U, z)}{dU} dE
+ \int_{0}^{eU} \frac{\rho_{tip}(eU - E)}{dU} \cdot \rho_{sample}(E) \cdot T(E, U, z) dE$$
(3.5)

Here $\rho_{tip}(E)$ is the tip DOS, $\rho_{sample}(E)$ is the sample DOS and T(E, U, z) is the transmission coefficient defined in [18]. T(E, U, z) depends on the tip-surface distance z and the effective barrier height, which itself is a function of E and U. To estimate T(eU, eU, z) one can make use of $I(U, z) \propto T(E = eU, U, z)$, which is valid as long as $U \ll \hat{\Phi}$. Measuring I(U, z) confirms the normally assumed exponential shape of T(E = eU, U, z)

$$T(E = eU, U, z) \propto \exp(-A \cdot \sqrt{(\hat{\Phi} - e|U|)/2} \cdot z)$$
(3.6)

with $A = \sqrt{8m}/\hbar$ (m: electron mass, \hbar : Planck's constant) and $\hat{\Phi}$ being the effective barrier height at $U = 0 \ mV$. As expected $\hat{\Phi}$ depends on the individual tip and varies between 1 eV and 2 eV.

Determining the Tip Induced Band Bending by Analyzing the Quantized States of the Tip Induced Quantum Dot

The vacuum potential between tip and surface differs by:

$$\Delta E_{Vac} = \Delta \Phi + eU \tag{3.7}$$

 $(E_{Vac}$: vacuum potential, Φ : work function). This in combination with the low charge density at the unpinned semiconductor surface leads to an electric field spreading into the sample and results in a tip induced band bending [19]. A quantitative interpretation of STS-data on InAs(110) requires an exact knowledge of the corresponding E_{CBM} -shift and the lateral extension of the band bending. At low temperatures, this information can be obtained by analyzing the quantized state energies of the tip induced quantum dot.

The electric field spreads in all three dimensions (3D) of the sample leading to confinement in 3D and resulting in a quantum dot (QD) centered below the tip. Its quantized states can be measured by STS. Comparison of the state energies with Hartree-calculations allows to reconstruct the potential of the QD [20].

The dI/dU-Spectra of the Tip Induced Quantum Dot

Fig.3.43a shows a typical dI/dU-spectrum measured in the region of the semiconductor band gap. The spectrum is recorded 50 nm apart from any dopant visible in topographic images. This avoids influences of the additional band bending of the dopant. A smooth curve is obtained except in the region between -0.2 V and 0.0 V. As expected from the DOS in the conduction and valence band of the sample, an increase of dI/dU is observed above 0.0 V and below -0.4 V. In fact the increase at negative

Figure 3.43: a.) dI/dU-spectrum of n-InAs(110); $U_{stab} = 200 \ mV$, $I_{stab} = 300 \ pA$; the bulk conduction band, the bulk valence band and the Fermi level (E_F) of the sample are indicated. The intensity is arbitrary but not shifted with respect to 0.0; b.) sketch of the tunneling gap and the band bending at $U \simeq 100 \ mV$; E_{BCBM} is the bulk conduction band minimum; the length of the arrows indicate different transmission coefficients; the tip-surface distance (6 Å) is arbitrary, but does not influence the shape of the tip induced band bending; c.) and d.) dI/dU-spectra of n-InAs(110); $U_{stab} = 100 \ mV$, $I_{stab} = 300 \ pA$, E_{BCBM} and E_F are indicated; the vertical lines mark the peak positions.

The spectra in a.), c.) and d.) are measured with three different, individually prepared W-tips.
voltages already starts at -0.2 V which will be explained below.

The dI/dU-signal does not vanish in the region of the band gap. To understand this, Fig.3.43b shows a sketch of the tunneling conditions at $U = 100 \ mV$. The Fermi level (E_F) of the degenerate InAs-sample is 12 meV above the bulk conduction band minimum (E_{BCBM}) . The tip induced band bending is downwards and induces quantized states ("tip induced states") into the band gap.

Both, E_{BCBM} and the tip induced states are located below E_F , i.e. $\rho_{sample}(E)$ vanishes only partly below E_F . Since T(E, U, z) and $\rho_{tip}(E)$ (metallic) are always larger than zero, the integrals of eq.3.5 would only vanish, if (over the whole integral) either dT/dUand $d\rho/dU$ vanish simultaneously or ρ_{sample} is zero. Both is impossible, since ρ_{sample} is already larger than zero at E_F and dT/dU is smoothly decreasing and positive in the whole band gap region. Consequently dI/dU(U) can not vanish in the band gap as found in the experiment. The decreasingly positive value of dT/dU directly explains the smoothly decreasing dI/dU-signal between -0.4 V and -0.2 V in Fig.3.43a.

The peaks between -0.2 V and 0.0 V are not explained so far. They are caused by the tip induced states⁴. Fig.3.43c and d show two spectra of the tip induced states in detail. They are measured with two microscopically different tips. The difference in the spectra indicates that the state energies depend on the individual tip.

The QD is the response of the sample to a tip charge, which is equivalent to a disk like (2D)-charge distribution in (x,y). Since the screening lengths of the sample is nearly isotropic, the QD is more extended in (x,y) than in z. Consequently, the ground state energy E_0 is dominated by the z-confinement.

We separate z from (x,y) leading to $E_0 = E_0^{x,y} + E_0^z$. The band bending in z can be estimated by twofold integration of the Poisson equation [19]. Assuming different $\Delta \Phi$, the potential V(z) is calculated and the ground state is fitted into the potential using a triangular potential approximation [21]. This procedure is repeated until the correct ground state energy is found. For Fig.3.43c the corresponding $\Delta \Phi$ is 250 meV, but for different tips it varies between 150 meV and 400 meV. The numbers are in agreement with the work functions of n-InAs(110) (4.9 eV [22]) and bulky W (4.55 eV [23]). Fig. 3.44a shows the calculated conduction band and the ground state wave function $\Psi(z)$ obtained for the dI/dU-curve in Fig.3.43c. To calculate the first excited state E_1^z one has to recognize that the bands are significantly curved. However assuming $dV/dz = \int dV/dz \, dz$ we can use the triangular approximation and estimate $E_{BCBM} - E_1^z = 5$ meV. This is indicated by the straight dashed line in Fig.3.44a. The change in ΔE_{Vac} with U is considered as indicated by the curved dashed line.

Obviously most of the peaks in Fig.3.43c can not be explained by z confinement. They are caused by (x,y) confinement. Their peak energies are nearly equidistant close to E_0 ($\Delta U_1 = 14 \ meV$) and decrease in separation at higher voltages ($\Delta U_2 = 8 \ meV$). Taking into account the changing ΔE_{Vac} and E_0 with U the corresponding energy distances are $\Delta E_1 = 8.8 \ meV$ and $\Delta E_2 = 5.2 \ meV$. A similar $\Delta E_1/\Delta E_2$ -relation is found in all experiments but ΔE_1 varies between 6 meV and 10 meV.

⁴A peak like $\rho_{tip}(E)$ restricted to a small energy region is improbable.



Figure 3.44: a) Shape of the tip induced band bending in z direction and the quantized state energies corresponding to the spectrum of Fig.3.43c (solid lines: U = -100 mV, dashed lines: U = -5 mV). The ground state wave function $\Psi(z)$ (Airy-function) is included [21]. The grey bars on the tip side at E_F indicate the energy region detected in dI/dU-curves at U = -110 mV and U = -5 mV; b) sketch of the lateral potential at U = -110 mV. + is the tip charge. The curved lines mark the electric field and the horizontal lines show the state energies. The ground state wave function $\Psi(x, y)$ is included.

Equidistant states imply a parabolic confinement [24]. However, electron-electron interaction has to be taken into account to analyze the state energies of the QD quantitatively. Therefore we performed Hartree-calculations with different trial potentials to reproduce the energy spectrum. The best agreement is found for a Gaussian external potential $V(r) = A \cdot \exp(-x^2/\sigma^2)$ with $\sigma = 60 \ nm$ and a parabolicity at the bottom of the QD corresponding to $\hbar\omega_0 = 17 \text{ meV}$.

Importantly, the parameters $\Delta \Phi$, σ and $\hbar \omega_0$ determine the potential below the tip and carry the necessary information for the correct interpretation of STS-data.

Magnetic Field Dependence of the Quantum Dot States

An appropriate test of the correctness of the model of the tip induced QD is its magnetic field dependence. It is well known that the n-th excited state of a 2D parabolic confinement is n-fold degenerate (neglecting spin degeneracy) and splits into n separate states in magnetic field [24]. At high magnetic fields the states bundle again and merge into Landau bands as observed in the absence of lateral confinement. This behaviour is calculated for the parabolic QD in the Hartree approximation⁵. The change of E_0^z with applied voltage is taken into account to compare the results (full lines in Fig.3.45a) with the experimentally observed peak positions (symbols in Fig.3.45b). The ground-state energy and the splitting of the first excited state is well reproduced by the calculation. This confirms our previous analysis of the tip induced

⁵The use of the Gaussian QD would imply that σ changes with applied voltage in a non-simple way. Since we evaluate only the lower lying states, we avoid this complication.



Figure 3.45: a.) dI/dU-spectra of n-InAs(110) measured at different B-fields perpendicular to the sample surface; $U_{stab} = 100 \text{ mV}$, $I_{stab} = 300 \text{ pA}$. The dotted lines connect shifting peaks of similar intensity (presumably states of identical quantum numbers). b.) peak positions as a function of magnetic field (symbols). Squares and circles represent the ground state and the first excited state of (x,y) quantization. Lines show the results of a Hartree calculation corrected for the changing ground state energy E_0^z with applied voltage U.

QD. The number of states at higher n (close to E_{BCBM}) is enormous and an analysis would require an increased energy resolution of the experiment.

The Influence of Dopants on the Tip Induced Quantum Dot

Since the QD moves with the scanned tip, it is possible to locate the QD above or close to *one* individual dopant and to study the influence of the dopant on the spectrum of the QD. The inset in Fig.3.46a shows a dI/dU-map of the sample with five visible and labelled dopants. They are located down to 10 nm below the surface. Only the visible dopants are within the z-FWHM of the ground state (GS) of the QD (10 nm). Their mean (x,y)-distance (50 nm) is larger than the (x,y)-FWHM of the GS of the QD (30 nm). Consequently, each dopant influences the GS of the QD individually.

dI/dU-Spectra of the GS of the QD taken above different dopants and a dopant free region of the sample are shown in Fig.3.46a. Above the dopants the GS is reduced in energy ($\Delta E = 5-8 \ meV$) and appears to be increased in intensity ($\Delta I/I = 25-45\%$). The differences between individual dopants are caused by their different depth z_D



Figure 3.46: a.) dI/dU-spectra taken at different positions of n-InAs(110); $U_{stab} = 50 \ mV$, $I_{stab} = 100 \ pA$. Spectra 1 - 5 are measured on top of the labelled dopants (bright dots) in the inset. The dopants are located at different depths up of 10 nm below the surface; spectrum 6 is averaged over the marked box; Inset: dI/dU-map, taken at $U = -55 \ mV$, 90 $nm \times 130 \ nm$; b.) squared wave functions $|\Psi(r)|^2$ of the calculated ground state of the QD, if a remote Coulomb potential is added (see text); z is the distance of the dopant from the 2D plane of the quantum dot. Energy shifts with respect to the unperturbed quantum dot are indicated. The used parameters (potential depth and dot diameter) fit the unperturbed spectrum.

below the surface.

Next, we compare the values ΔE and $\Delta I/I$ with Hartree-calculations by adding a remote Coulomb potential $V_{Coul}(r)$ to the calculated 2D-QD:

$$V_{Coul}(r) = \frac{e}{4\pi\epsilon\epsilon_0\sqrt{r^2 + z^2}}$$
(3.8)

 $(\epsilon = 14.8 \ (\epsilon_0)$: dielectric constant of InAs (vacuum), r: (x,y) distance from the dopant, z: z distance of the dopant from the plane of the QD).

The resulting wave functions $\Psi(r)$ of the GS's and their corresponding energy reductions $\Delta E(z)$ are shown in Fig. 3.46b for different z. To consider the finite extension of the QD in z and the different z_D we used the unperturbed Airy-function of Fig.3.44a as a probability distribution $\Psi(z)$. It follows: $\Delta E(z_D) = \int \Delta E(z - z_D) \cdot \Psi(z) dz$ [13]. $\Delta I/I(z_D)$ is calculated analogously. The results are $\Delta E(z_D) = 9 - 14 \ meV$ and $\Delta I/I(z_D) = 20 - 80\%$. If one takes into account that the measurement above the dopant free surface is influenced by a background potential of about $-3 \ meV$, both values are in good agreement with the experimental data.

The evolution of the GS, if the QD is moved away from the dopant, is shown in Fig.3.47a. As expected the energy reduction decreases with increasing distance from the dopant. But, interestingly, the intensity appears to have a minimum about 12 nm away from the dopant. This is explained qualitatively by the sketch in Fig.3.47b. The GS sticks at the dopant and the intensity in the center of the QD can be reduced below its value without dopant.



Figure 3.47: a.) dI/dU-spectra taken at different positions with respect to an ionized dopant; $U_{stab} = 50 \ mV$, $I_{stab} = 100 \ pA$. The spectra are measured (left to right) 0 nm, 4 nm, 8 nm, 12 nm, 16 nm and 30 nm apart from the dopant as indicated in the inset. Thick lines correspond to the ground states sketched in b.); Inset: dI/dU-map, taken at $U = -55 \ mV$, 34 $nm \times 34 \ nm$; b.) sketch of the ground state of the unperturbed dot (solid lines) and the dot perturbed by an off-center dopant (dashed lines). The dashed line represents the z-averaged ground state calculated for a dopant lying 4 nm below the surface. I_t indicates the tunneling position.

Scattering Electron Waves at Ionized Dopants



Figure 3.48: a.) dI/dU-image of n-InAs(110): 200 $nm \times 200 nm$, U = 50 mV, I = 400 pA, $U_{mod} = 8.5 mV_{pp}$; b.) constant current image of the same surface area as a.); crosses mark corresponding positions.

Inhomogenities in the crystal potential lead to the scattering of electrons. The resulting interference pattern of the electron waves can be visualized by STM [25]. The attractive potential of dopants is an inhomogenity of the potential and dI/dU-images show the corresponding scattering states [26].

Fig.3.48a shows a dI/dU-image recorded in constant current mode in comparison with a topographic image of the same area (Fig.3.48b). A number of concentric ring structures is visible in Fig.3.48a. There are more concentric ring structures in Fig.3.48a than elevations in Fig.3.48b. However, if an elevation marks the position of a dopant in Fig.3.48b, it has the same center as a ring structure in Fig.3.48a. The different radii of different ring structures are caused by the different z_D of the dopants. Increasing the applied voltage U and therefore the energy of the detected electronic states reduces the diameter of the rings as expected and is shown in Fig.3.49 a1 – a7.

dI/dU(U, x, y)-images depend on $\rho_{sample}(eU, x, y)$ and T(E = eU, U, z(x, y)) (see eq.



Figure 3.49: a1.) - a7.) dI/dU-images measured at the sample voltages indicated above, 100 $nm \times 100 nm$, I = 400 pA, $U_{mod} = 8.5 mV_{pp}$; b1.) - b7.) (dI/dU)/I(z)images corresponding to a1.) - a7.); c1.) - c7.) simulated surface distribution of scattering states of an ionized dopant ($z_D = 14.3 nm$); state energies correspond to the voltages above (see text); d1.) - d7.) Comparison of circular line sections of b1.) - b7.) and c1.) - c7.); a circular line section represents the average intensity on a circle of constant distance from the center of the rings; smoother curves belong to the simulated images.

 $(3.5)^6$. To measure $\rho_{sample}(E, x, y)$, the influence of T(E = eU, U, z(x, y)) has to be eliminated. This is done by measuring I(z) with each individual tip after recording dI/dU-images. Using eq.3.6 and the fact that z(x, y) (topographic image) is measured in parallel with dI/dU(x, y) we can rearrange eq.3.5:

$$\rho_{sample}(E, x, y) \propto \frac{dI/dU(U, x, y)}{I(U, z(x, y))}$$
(3.9)

By inserting the measured data one gets directly the LDOS of the sample. Fig.3.49 b1 – b7 shows the ρ_{sample} -images corresponding to a1 – a7.

The measured $\rho_{sample}(x, y, E)$ can be compared to the expected LDOS of the scattering states of ionized dopants. To calculate the scattering states we used a one dimensional WKB-model [27]. The potential consists of the attractive dopant potential $V(r) = e/(4\pi\epsilon\epsilon_0 r)$ (r: distance from the dopant)⁷ and the tip induced band bending known from the analysis on page 67 ff.. Fig.3.50 shows a 3D-sketch of the potential $E_{CBM}(\vec{r})$. From the potential the squared wave function of the scattering state is constructed as usual [27]. A local wavelength $\lambda(E - E_{CBM}(r))$ is calculated with the help of the known dispersion of the InAs-conduction band [28]. Phase co-

⁷Screening effects are of minor importance and can be neglected.



Figure 3.50: 3D-image: calculated potential caused by an ionized dopant $(z_D = 14.3 nm)$ and the tip induced band bending $(\Delta \Phi = 160 meV)$; the lateral extension (x,y) of the tip induced quantum dot is not visible on that length scale; 2D-image: $\sin^2(\varphi)$ -distribution of the resulting scattering state at $E - E_F = 50 meV$ (see text for discussion).

⁶The second and third term of eq.3.5 can be neglected at low U. An analysis shows that they cause much less than 10 % of the lateral corrugations in dI/dU(x,y) at $U \leq 200 \ mV$.

herence is preserved by integrating a phase factor $\varphi(\lambda(r))$ stepwise starting from the center of the dopant. The bottom of Fig.3.50 shows the resulting $\sin^2(\varphi(r))$. The squared wave function of the scattering state in 3D is proportional to $1/r^2 \cdot \sin^2(\varphi(r))$. Its value directly below the tip corresponds to $\rho_{sample}(x, y, E)$. It has to be calculated separately for each (x, y) and E.

The important point is, that the simplified model leaves only *one* parameter to fit size, energy dependence and intensity relations of the concentric ring structures of each dopant: namely z_D . The best fit for Fig.3.49 b1 – b7 is achieved for $z_D = 14.3 nm$. It is shown in Fig.3.49 c1 – c7. For a quantitative comparison Fig.3.49 d1 – d7 show circular line sections of the measured and the calculated images in comparison.

To check the correctness of the fit parameter z_D the described procedure is applied to the 10 best visible structures in Fig.3.48b (200 $nm \times 200 nm$). The 10 dopants appear to be distributed in a z-slice of 20 nm implying $N_D = 1.25 \times 10^{16}/cm^3$. The difference to the measured dopant concentration ($2.0 \times 10^{16}/cm^3$) is caused by the fact, that some of the patterns could not be analyzed. They overlap too strongly with neighbouring patterns. Indeed about 15 ring structures are visible in Fig.3.48b, but only 10 are evaluated.

An exact knowledge of z_D would allow to determine reflective scattering cross sections $\sigma_R(E)$ of individual dopants. Therfore we compare the amplitude of the corrugations with the total signal of $\rho_{sample}(x, y)$. But even without the exact knowledge of z_D , $\sigma_R(E)$ can be given in arbitrary units. This is shown in Fig.3.51 for the dopant analyzed in Fig.3.49. A trend of decreasing σ_R with decreasing E in qualitative agreement with scattering theory is found. The data are compatible with a $1/E^2$ -fit as expected from the Born-approximation, but a 1/E-fit seems to be more appropriate. Further investigations are necessary to determine $\sigma_R(E)$ in more detail.



Figure 3.51: Reflective scattering cross section $\sigma_R(E)$ of the dopant analyzed in Fig.3.49. Assuming a depth of the dopant $z_D = 14.3 \ nm$, the y-scale would be in nm^2 . The lines show 1/E- and $1/E^2$ -curves for comparison.

Landau Quantization

InAs exhibits an isotropic bulk conduction band close to BCBM and consequently an isotropic 3D electron gas. In magnetic field (B) its square root like DOS splits into Landau bands. The energy separation of the Landau bands is $\Delta E = \hbar e B/m_{eff}(E)$. An approximation of the increasing $m_{eff}(E)$ with E, reflecting the non-parabolicity of the InAs conduction band, can be given by the $k \bullet p$ -theory [28]:

$$m_{eff}(E) = m_{eff}(E_{BCBM}) \cdot (1 + \frac{2(E - E_{BCBM})}{E_{Gap}})$$
(3.10)

 $(E_{Gap} = 0.4 \ eV: \text{ band gap}, m_{eff}(E_{BCBM}) = 0.023 \times m_e, m_e: \text{ electron mass})$



Figure 3.52: a.) dI/dU(U)-curves taken on a dopant free area of the surface at different magnetic fields perpendicular to the surface; $U_{stab} = 100 \ mV$, $I_{stab} = 300 \ pA$, $U_{mod} = 3 \ mV_{pp}$; b.) peak positions of the oscillations in a.); symbols (+) and (×) mark peaks measured with the magnetic field in opposite directions.

The Landau quantization is detectable in dI/dU(U)-curves taken in magnetic fields up to 6 T perpendicular to the sample surface [29]. Fig.3.52a shows the corresponding data taken on a dopant free region of n-InAs(110). The peaks below 0 mV correspond to the tip induced states analyzed on page 67 ff. The spectral part above 0 mV belongs to the DOS of the bulk conduction band. The oscillations at B > 2 T are due to the Landau quantization. Determining the peak positions results in Fig.3.52b. Obviously the peaks can not be connected by straight lines indicating that $m_{eff}(E)$ is not constant. The m_{eff} -data extracted from the energetic distances of the peaks are shown in Fig.3.53 (open circles) together with the results of eq.3.10 (dashed line) and data from D.C. Tsui obtained by analyzing the transport properties of planar tunnel junctions (filled dots) [30,31]. The agreement confirms that STS is able to detect the energy dependence of m_{eff} . Smaller deviations of the linear regression line (full line) from the results of eq.3.10 are not significant as checked by analyzing different data sets.



Figure 3.53: Effective mass data as determined from Fig.3.52 (open circles). One typical error bar is marked; solid line: linear regression; dashed line: effective mass expected from $k \bullet p$ -theory [28]; black dots: effective mass data from [30].

Local Fluctuations of the Landau Energies

The importance of the measurements described in the last section is its implied possibility to measure Landau energies with high spatial resolution. The ionized dopants of the sample lead to potential fluctuations which influence the Landau energies. The most simple way to estimate the energy shifts is a perturbation theory, strictly valid



Figure 3.54: dI/dU-curves recorded at 6 T on a dopant free area (dashed line) and at the position of a visible dopant (solid line); $U_{stab} = 100 \ mV$, $I_{stab} = 300 \ pA$, $U_{mod} = 3 \ mV_{pp}$; Spectra are averaged over regions of 10 $nm \times 10 \ nm$. LL-numbers and peak shifts are indicated.

for a one electron model in 2D. It results in energy shifts:

$$\Delta E_n = \langle \Psi_n | V(\vec{r}) | \Psi_n \rangle . \tag{3.11}$$

 Ψ_n is the wave function of the n-th Landau state having a radial contribution of $\Psi_n(r) \propto r^n \cdot \exp(-r^2/d_0^2)$ [32]. The parameter $d_0 = \sqrt{\hbar/(eB)}$ is the magnetic length being 21 nm for 6 T. The local potential $V(\vec{r})$ is provided by the ionized dopants. The model implies that the ring shaped Landau state simply tests the local potential averaged over $\Psi(r)$.

Including the third dimension into the calculation would lead to a mixing of different Landau levels (LL's) and a reduction of the energy shift of the lower LL's [33]. The inclusion of more electrons into the model would lead to the problems of screening in magnetic field. Clearly, a quantitative comparison would require an unrestricted Hartree-Fock-calculation [34].

However, to estimate the expected energy shifts we consider eq.3.11 with the two extreme cases of a fully unscreened Coulomb potential $V_{Coul}(r)$ (eq.3.8) and the screened Coulomb potential as expected at B = 0 T. The last is obtained by multiplying eq.3.8 with $e^{-r/\lambda}$ ($\lambda = 16 \ nm$) [16]. The resulting ΔE_n for n = 0 - 9 are as follows: screened case: 3.6 meV/1.8 meV/1.1 meV/0.9 meV/.../0.6 meV;

unscreened case: 9.5 meV/5.5 meV/4.3 meV/3.6 meV/.../2.2 meV.

Fig.3.54 shows two dI/dU(U)-curves. The dashed one is obtained above a dopant free area of the surface and the other, solid one is obtained directly above a dopant. Energy distances of the LL's are indicated. They correspond to the estimates given above.

One recognizes immediately from Fig. 3.54 that the energy shifts should result in large fluctuations of $\rho_{sample}(E, x, y)$. This can be visualized in dI/dU-images. At 50



Figure 3.55: n-InAs(110), $U = 50 \ mV$, $I = 400 \ pA$, $500 \ nm \times 500 \ nm$; a.) $B = 0 \ T$, constant current image; b.) $B = 0 \ T$, dI/dU-image; c.) $B = 6 \ T$, dI/dU-image. White lines mark identical regions in a.), b.) and c.).

meV the LDOS above dopant free areas should be dominant and indeed Fig.3.55c shows bright regions where few dopants can be detected in the topographic image (Fig.3.55a) and the dI/dU-image recorded at B = 0 T (Fig.3.54b).

Fe on InAs(110)

As mentioned in the introduction, InAs(110) is chosen for the project, since it builds up an adsorbate induced 2DEG at the surface. Future experiments will investigate the interaction of the 2DEG with ferromagnets and superconductors. Here, we will present first results concerning the system ferromagnet/2DEG.

Fe on InAs(110) deposited at room temperature pins E_F in the bulk conduction band. This is demonstrated by taking UPS-spectra of the p-InAs(110)-valence band at different Fe-coverages. An example is shown in Fig.3.56a. Obviously, the spectral shape is nearly unchanged at different Fe-coverages but the whole valence band (VB) is moved down with respect to E_F . A quantitative evaluation of the corresponding upwards E_F -shift with respect to the VB-maximum is shown in Fig.3.56b. At a coverage of 2% Fe E_F is pinned in the CB.

Fig.3.57a shows an STM-image of n-InAs(110) covered with 5 % Fe. The Fe-atoms are arranged into 9 monomers, 6 dimers, 3 trimers and 1 tetramer. This arrangement indicates that Fe-atoms are mobile at room temperature⁸. Since there are no indications of exchange processes, it appears to be possible to prepare ferromagnetic Fe-islands $(r \simeq 6 nm [35])$ with atomically sharp interfaces, which would be an ideal model system of a ferromagnet/2DEG interface. Fig.3.57b and c show larger scale STM-images of 4 %Fe on n-InAs(110) and, for comparison, the clean n-InAs(110)-surface. Large

 $^{8}\mbox{After Fe-deposition}$ the sample was held at room temperature for 5 min before it was cooled down to 6 K.



Figure 3.56: UPS-data of Fe/p-InAs(110) ($N_A = 3 \times 10^{18}/cm^3$): a.) valence band spectrum recorded at $h\nu = 15 \ eV$, $\theta = 30^{\circ}$ and indicated Fe-coverages; b.) resulting E_F -shift as a function of Fe-coverage.



Figure 3.57: Constant current images of Fe/n-InAs(110): a.) 5 % Fe, U = 50 mV, I = 100 pA, 12 $nm \times 12 nm$; b.) 4 % Fe, U = 150 mV, I = 200 pA, 50 $nm \times 50 nm$; c.) 0 % Fe, U = 50 mV, I = 300 pA, 50 $nm \times 50 nm$

scale contrasts are induced by the Fe (Fig.3.57b). The pattern indicates the complexity of the build-up of the 2DEG at the surface, but also, and more importantly, the ability of the STM to study the evolution of the 2DEG in detail.

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3.3.2 Scanning Capacitance Microscopy and Spectroscopy Applied to Semiconductor Devices

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The activities of the second nanoelectronic subgroup are divided into two sections: the characterization of semiconductor devices and the fabrication of structures in the nanometer regime. At first, we present the methods for the analysis of integrated circuits.

Characterization of Nanostructures by Means of SPM Methods

With decreasing size of integrated circuits (ICs), there is an increasing demand for the measurements in the sub-micrometer regime, and most of the state-of-the-art analytical tools failed in this regime. So, we need new powerful methods for the analysis of semiconductor devices with a lateral resolution of less than 100 nm. On the other hand, scanning probe microscopes (SPMs) are predestined for this demand, due to their twodimensional (2D) imaging capabilities, high spatial resolution, and non-destructive nature. The whole family of SPMs offers the detection of each significant value (Fig. 3.58).

We have shown, that the scanning thermal microscope (SThM) allows the detection of hot spots on the IC with resolution in the sub-micrometer regime. By means of the magnetic force microscope (MFM), we are able to detect the magnetic stray field of a conductive wire with diameter of less than 200 nm. A calculation gives us the current of the conductor. Hence, we have shown the measurement of current of some tens of μ A with a lateral resolution of less than 200 nm on an IC. The time resolution of this method, unfortunately, is limited due to the resonance frequency of the cantilever used on the order of some tens of kHz. On the other hand, one can use other SPM methods, for example the scanning Hall probe microscope (SHPM), to overcome this limit. An electrostatic force microscope (EFM) enables the measurement of an electric field and therefore a voltage on an IC, with spatial resolution of less than 100 nm. Furthermore, special techniques (heterodyne mixing) enable the measurement of voltage with frequencies of some tens of GHz. By means of the scanning nearfield optical microscope (SNOM or NSOM), one can explore the optical properties of semiconductor devices, for example laser diodes, in the sub-micron regime.

An important point is the detection of 2D-doping profiles with lateral resolution of less than 100 nm and dopant concentration of 10^{15} to 10^{20} cm⁻³. The "National Technology Roadmap for Semiconductor" describes this measurement as decisive technology for the next generation of ICs. This method would permit the determination of the effective channel length and the lightly-doped drain (LDD). Exact data are very important for the optimization and calibration of technology computer aided design



Figure 3.58: An overview of useful SPMs for the analysis of integrated circuits.

(TCAD) systems, and new designs of MOS-FETs and Bipolar-transistors require an exact knowledge of the 2D-doping profiles. Today three methods offer the potential to map 2D-doping profiles: the Kelvin force probe microscope (KFM) measures the surface potential [1], the scanning spreading resistance profilometer (SSRP) [2] measures the resistance profile of the sample and requires a good electrical contact between probe and sample, whereas the scanning capacitance microscope (SCM) measures the capacitance profile and requires a thin oxide layer [3]. Consequently the disadvantage of the second method is the breakdown of the thin natural oxide due to the use of high forces of up to 100 μ N, so the sample is damaged and the measurement is not repeatable. Furthermore we have problems to find a suitable cantilever and to prepare the sample for this application, due to the demand for an uniform doping concentration between the back contact and the ohmic contact of the tip. The disadvantage of the SCM is its sensitivity to often unknown parameters of the insulator (thickness and dielectric constant) which makes a quantitative analysis more difficult. However, the manufacturing of high quality thin oxide layers is a standard procedure in semiconductor processing and often there is only a need for the knowledge of the exact position of a junction.

The KFM has a worse spatial resolution compared with the SCM (Fig. 3.59). This



Figure 3.59: A measurement of the same location of a test structure by (a) SCM and (b) KFM is taken. The SCM gives us a better lateral and dopant resolution.

is due to the long-range electrostatic interaction between tip and sample. Thus only 50% of the tip force of the KFM signal acts on the tip within a borderline of 100 nm. Therefore we take some further investigation of the methods with the higher lateral and dopant resolution based on the SCM.

The SCM is a scanning force microscope (SFM) with a conductive probe connected to a capacitance sensor. This sensor is based on the method of a resonating circuit (Fig. 3.60a). The signal f_g of a highly stabilized oscillator of nearly 1 GHz is coupled in a bandpass with a resonating frequency of f_0 . The signal is coupled out and rectified. A change of the resonating frequency of $\Delta f_0 = /f_0 - f_{0'}/$ results in a change of the output voltage $\Delta U_{out} = /U_{out} - U_{out'}/$ (Fig. 3.60b). The sensor enables the detection of a capacitance of less than 1 aF (10⁻¹⁸ F). Furthermore we used lock-in technique



Figure 3.60: (a) Experimental setup and (b) measurement principle of the SCM.

to avoid the influence of stray capacitance. By sweeping the bias voltage V_{bias} and by observing the output voltage, we get a capacitance-voltage (C(U)) curve of a metaloxide-semiconductor (MOS) system. Here we have to distinguish three parts: the accumulation, depletion and inversion parts (Fig. 3.61a). Within the accumulation region, there is a change of the majority carrier at the silicon-oxide interface. With increasing voltage the MOS system creates a depletion region, so that the change of the majority carriers reaches deeper into the silicon, depending on the doping level. Due to band bending, the MOS system has the potential to build up an inversion layer of minority carriers at higher positive voltages (Fig. 3.61a; LF: low-frequency curve). However, the change of the density of the minority carriers has a large time constant, therefore the minority carriers cannot follow a driving signal with a frequency of more than 1 kHz. Because of the high frequency of the measurement signal and a sweep rate of the bias voltage on the order of some V/s, we should measure only the dC/dU(U) deep depletion curve.



Figure 3.61: (a) A theoretical C(U) curve (LF: low-frequency curve, DD: deep depletion curve) (b) two dC/dU(U) curves for different doping levels.

Lanyi et al. [4] used another capacitance sensor with a lower driving frequency of some 100 kHz and active shielding. Furthermore, they used their SCM to measure the topography of a metallized sample, but they could not measure the topography and the capacitance signal simultaneously and independently. Lanyi et al. [5] calculated for this setup a limit of the lateral resolution of a few nanometer. Our setup measures the change of carriers in a doped silicon. Therefore the lateral resolution depends on the doping level . To overcome this disadvantage, we used the bias feedback mode developed by Williams et al. [6]. This mode changes the value of the AC bias voltage (Fig. 3.61 $V_{mod,\omega}$, ω typically 100 kHz) and keeps the size of the change of the depletion region constant. So we are able to reach a lateral resolution of some 10 nm depending on the tip radius. Fig. 3.61b shows the differentiated C(U) curve for different doping levels. At 0 V bias the signal of the lower doped sample is higher than the signal of



Figure 3.62: A comparison of a C(U) curve taken by a silicon cantilever and a metal coated silicon cantilever on a silicon sample is given.

the higher doped sample, but at 1 V bias it is reversed. Hence, the bias voltage is an important factor in SCM, and a change of the bias gives us additional information about the sample.

To test the influence of the bias voltage on the topography signal, the bias voltage was sweeped with an amplitude of 10 V and a sweep rate of 5 Hz during the scan. A spectrum analysis of our measured signals gave us a distinct peak in the capacitance image, but there was no significant peak at 5 Hz, neither in the topography nor in the error signal of the force feedback. Furthermore, we applied a 50 V pulse (100 ms) on a 50 nm thick oxide (corresponding to an electrical field of 10^7 V/cm), but saw no influence on the surface. That means, that we found no significant influence of the bias voltage on the stability of the topography signal and the oxide layer, for oxide thicknesses of more than 3 nm. The conductive probes used are silicon cantilevers with cobalt/chromium coating. One can use any other cantilever, for example (doped) silicon or (doped) diamond cantilevers, but the influence of this doped cantilever must be taken into account (Fig. 3.62). These cantilevers have typical tip radii of less than 40 nm and a resonance frequency of 50 to 80 kHz.

Our standard test sample is n-doped silicon $(10^{16} \text{ cm}^{-3})$ with highly p-doped strips $(10^{19} \text{ cm}^{-3})$. The topography of Fig. 3.63a shows remainders of the doping mask, and the light bright stripes of the capacitance image show the lower n-doped regions (Fig. 3.63b). The width of the doped stripes (dark lines in the capacitance image, Fig. 3.63b) are broader than the space between the remainders of the mask (dark stripes in the topography, Fig. 3.63a).

After a scan of $30 \times 30 \ \mu m^2$ with 0 V bias and a scan rate of 1 Hz, we scanned an area of $1 \times 1 \ \mu m^2$ with 5 V bias in the middle of the previous scan frame. Then the same $30 \times 30 \ \mu m^2$ area was scanned with 0 V bias. Fig. 3.63d shows clearly the influence of previous measurements on subsequent ones. The following images taken



Figure 3.63: (a) Topography and (b) capacitance images with 0 V bias of a virgin area of the sample (c) Topography and (d) capacitance images with 0 V bias of the same area, after a 1 $\mu m \times 1 \mu m$ scan with 5 V bias inside the imaged area has been performed.

at the same location showed further spreading of the influence spot. After half an hour the spot is vanished. Furthermore, we observed low-frequency curves with our high-frequency equipment (Fig. 3.64a). Our explanation for all these phenomena is based on surface charges, which have the potential to form a permanent inversion layer underneath the surroundings of the probe.

Fig. 3.64c represents a model for a measured C(U) curve (Fig. 3.64b). At -3 V bias we can see the charging and discharging of the inversion layer underneath the probe by means of the electrons (minority carriers at p-silicon) from the surroundings. This explains the dependence of SCM-measurements on environmental conditions. We observed qualitatively this dependence due to the change of relative humidity. Hence, it is an important condition in SCM to avoid surface charges or ions. Thus we need a good sample preparation method. A heat treatment of the sample of 150° C for thirty minutes reduces significantly the influence of surface ions.



Figure 3.64: (a)-(c) Measured dC/dU(U) sweeps from accumulation (positive voltage) to inversion (negative voltage). (b) Integrated curve of (a) shows low-frequency behavior (Fig. 3.61). (c) Model for the explanation of curve (b).

In addition, we developed a new method, Scanning Capacitance Spectroscopy (SCS), to measure an array of dC/dU(U) curves. This gives us additional information about the sample. Therefore, we are able to distinguish between points on a sample, even where we cannot see any difference due to a capacitance line scan with a constant bias voltage (Fig. 3.65). Only an array of dC/dU(U) curves gives us the whole capacitance information of the sample. This knowledge enables us to use SCM as a powerful tool for the analysis of integrated circuits.

To convert the measured signal (dC/dU) into a doping profile we need the dC/dU dependence on the doping concentration. This connection is made with numerical calculations by means of a device simulator (DESSIS) [7]. First, we calculated 1D structures, to check the correctness of the result of the simulator compared with the analytical solution. Following, we simulated 2D (cartesian coordinates) and next 3D (cylindrical coordinates) geometries (Fig. 3.66).



Figure 3.65: (a) Measured topography and (b) dC/dU line scan with 0V bias. Points 2 and 15 of (b) show the same value. Therefore, we cannot distinguish between these points of the sample, due to a dC/dU measurement at a fixed bias voltage. On the other hand, the dC/dU(U) curves taken from these points on the sample ((c) number 2 and (d) number 15) exhibit a great difference.

With a realistic tip-sample geometry, we calculated C(U) curves for different homogeneous doping concentrations and determined the dC/dU dependence on the doping concentration (Fig. 3.67). Thus we are able to convert the SCM signal into a 2D doping profile, by means of a doping standard.

In Fig. 3.68 an example is presented showing that the capacitance image complements the topography of an integrated circuit. The sample is a cut of a DRAM and shows the area of the refresh transistors. The SCM (contact mode) offers the exact measurements of the gate strips of these transistors.

In addition, we take some investigations of electrostatic discharges (ESDs). ESD structures are designed to protect the device from high-voltage peaks that may destroy parts of the IC. Analysis of device failures due to breakdown phenomena requires the investigation of low currents directly or the interaction products resulting from a



Figure 3.66: Simulated space charge density for a low-doped Si-sample in depletion, calculated in cylindrical coordinates by means of the device simulator ISE-DESSIS.



Figure 3.67: Example for an inversion curve for measurement data, calculated in cylindrical coordinates by means of the device simulator ISE-DESSIS.



Figure 3.68: (a) Topography and (b) capacitance images of a $(30 \times 30 \ \mu m^2)$ area of refresh transistors of a DRAM.

current, for example the change of the dopant profile.

The topography of a SCM image of a EDS structure (Fig. 3.69a (after background subtraction)) indicates the channel of the electrostatic breakthrough. The capacitance



Figure 3.69: (a) Topography and (b) capacitance images of the ESD structure.



Figure 3.70: (a) Topography and (b) capacitance images of a NOS heterostructure with stored charged dots.

measurement (Fig. 3.69b) shows a fairly low contrast in the channel area and a disturbed p-well where the punch-through had taken place. This led to a uniformly doped area which enabled current to pass through. Furthermore we demonstrated the quick locating of a pn-juntion by means of SCM, without a time-consuming sample preparation (only a HF dip was carried out). To summarize, we have shown the usefulness of qualitative SCM as a analytical tool for the failure analysis now, and have taken some steps toward quantitative SCM in the near future.

Fabrication of Nanostructures by Means of SPM Methods

Another interesting field of application for SPM-based technologies are ultra high density storage (UHDS) devices. Up to date hard disks reach data densities of less than 2 Gbits/in², and the border line for conventional storage technology are on the order of several tens of Gbits/in² because of limitations regarding superparamagnetics in magnetic recording and optical diffraction in optical recording. But nevertheless, there is an increasing demand for higher data capacities. By using a STM, one can manipulate single atoms and reach a data density of 10^6 Gbits/in². The disadvantages are the great technical effort and the slow scan speed and data transfer rate respectively. To use other SPMs seems to be a better approach.

Gordon Kino and colleagues at Stanford University [8] developed a solid immersion lens scanning near-field optical microscope (SIL-SNOM). With this technique the IBM group of Bruce Terris [9] achieved a data density of 2.5 Gbits/in² and a data rate of 3.3. The same group presented a read-only-memory (ROM) device with a data density of more than 60 Gbits/in² [10]. They used a polycarbonate disk with smallest pits of 100 nm, and a SFM for read out. The disk was manufactured from an e-beam generated master pattern. This technology enables mass production of samples with pits of less than 50 nm. The SFM used works in contact mode with special high-frequency piezoresistive cantilevers having resonance frequencies of up to 10 MHz. Furthermore, Terris et al. have developed a SFM-based system with a rotating sample. A new autotracking system could maintain the tip on a particular data track. Thus they get a data density of 65 Gbits/in² and a data rate of more than 1 Mbits/s.



Figure 3.71: Schematic of the ROD-device developed at MARCH.

A possible random access memory (RAM) UHDS device is based on a SCM and on charge storage in a nitride-oxide-silicon (NOS) heterostructure (Fig. 3.70). A voltage pulse of typically 40 V and 10 μ s allows the charging of the nitride-oxide interface, and a voltage pulse of the opposite direction can discharge this interface. The smallest structure written in our laboratory had a full width at half maximum (FWHM) of 160 nm. The maximum data density reached is more than 30 times higher than in todays commercial storage memories. The theoretical data density limit determined by the overlap of the depletion areas is more than a hundred times higher. To reach high data rates, a SCM-based prototype (rotating disk (ROD) -device) has been developed that rotates the sample (Fig. 3.71). The SCM head can be moved by a high-power piezoelectric element in z-direction, so that the cantilever is pressing on the surface with a force as small as possible. First tests gave us data rates of more than 100 kHz, with a signal to noise ratio of 60 dB. The great problem of this storage technology is the contact between the metallic tip of the cantilever and the very hard silicon-nitride surface. At relative velocities of some meter per second, the tip gets rubbed off and the resolution of the system and its data density decreases as a function of time. This wear problem can be expected to be diminished by development of new coating technologies or non-contact measurements. A break-through could result from investigations in nanotribology by means of friction force microscopy (FFM), see section 3.4.2. To summarize, our work shows the possibility to use SPM-based techniques for a future UHDS device.

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3.4 Nanomechanics

From the broad range of possible applications, the nanomechanics subgroup focused mainly on two different topics during the past three years: First, we studied the forces which are connected with the relative motion of two bodies in contact. The effects associated with these forces are well-known to everybody as friction, and the branch of science dealing with these effects at the nanometer scale is called *nanotribology* (see Sect. 3.4.1). Second, the interactions between a sharp tip and a sample surface close before contact or after contact are used to obtain spatial information on the tip-sample interaction potential and the local elasticity with high resolution. The technique which is applied is called *dynamic scanning force microscopy* and will be explained further in section 3.4.2.

3.4.1 Nanotribology

U. D. Schwarz, H. Bluhm, H. Hölscher, and O. Zwörner

Introduction

Frictional forces are familiar from daily life; their understanding is important for the optimization of all kinds of machinery with parts in relative motion. Consequently, *tribology*, the science of friction, wear and lubrication, has a long history [1]. Astonishingly, only little progress was made in finding an exact physical description of friction and its associated phenomena since nearly 200 years ago, the following phenomeno-logical laws were found by Amontons [laws (i) and (ii)] and Coulomb [law (iii)]:

- (i) Friction is *proportional* to the load (i.e. to the applied normal force).
- (ii) Friction is *independent* of the (apparent) contact area.
- (iii) Sliding friction is *independent* of the sliding velocity.

These laws apply surprisingly well for *dry friction* (i.e. in the case of friction without lubrication, which is often also called *Coulomb friction*). Nevertheless, they could not be derived from (microscopic) principles up to now. Moreover, the understanding of the fundamental mechanisms of friction on the atomic level is generally poor since most macroscopically measurable frictional effects are dominated by the influence of wear, plastic deformation, lubrication, surface roughness, and surface asperities.

In recent years, however, the field of *nanotribology* was established by introducing new experimental tools, mainly the quartz crystal microbalance [2-4], the surface force apparatus (SFA) [5-7], and the scanning force microscope (SFM) [8-10], which allow the investigation of wearless friction on the nanometer scale. Today, macroscopic friction is believed to be the sum of the frictional effects occurring at many individual small asperities which make the physical contact between the two bodies sliding relative to each other. Thus, the investigation of "point-contact friction", experimentally realized in an SFM by a single-asperity contact of a sharp tip with a suitable sample, seems to be especially promising.

Since the pioneering work of Mate *et al.* [8] who succeeded 1987 in detecting frictional forces with the scanning force microscope and correlating them with atomic-scale processes, scanning force microscopy studies did indeed contribute significantly to the understanding of friction on the nanometer scale. Surprisingly, friction manifests itself at the nanometer scale in a significantly different manner than the Coulomb friction does at the macroscopic scale. Therefore, one of the main topics of the nanomechanics subgroup in the years 1996-98 was the investigation and the analysis of the frictional behavior of materials in dry friction at the nanometer or even at the atomic scale; the results of this work will be described in the following. For example, we found that the friction laws (i) and (ii) are not valid on the nanometer scale, since the frictional forces are proportional to the *actual* area of contact, which is generally not proportional to the load [11, 12]. On the other hand, friction law (iii) could be confirmed also on this scale [13].

The Scanning Force Microscope as a Tool for Nanotribology

In a scanning force microscope (see Fig. 3.72a), a sharp tip which is mounted at the end of a very soft leaf spring (the *cantilever*) profiles the surface of a suitable sample. During data acquisition, the sample is scanned in x-y-direction by means of piezoelectric elements, and the deflection of the cantilever is determined simultaneously with (ideally) sub-Ångström precision. A more detailed description of the method and the generally used instrumentation can be found, e.g., in Ref. [16].

The experimental set-up for the deflection detection measurement, which is most frequently used in SFM studies on friction, is the beam deflection scheme sketched in Fig. 3.72a. With this detection scheme, not only the deflection, but also the torsion of the cantilever can be measured (see the caption of Fig. 3.72a for further details). Force microscopes, which can record bending and torsion of the cantilever simultaneously, are often referred to as friction force microscopes (FFMs) or lateral force microscopes (LFMs). During scanning, the torsion of the cantilever is exclusively induced by lateral forces acting in x-direction (F_x) , whereas for corrugated samples, the bending of the cantilever is mainly caused by forces acting perpendicular to the sample surface in z-direction (F_z) . Nevertheless, a bending of the cantilever induced by lateral forces acting in y-direction (F_y) cannot be excluded (see Fig. 3.72b bottom). However, for atomically flat surfaces, Fujisawa et al. [14, 15] pointed out that the variation of F_z is so small that the measured signal of an SFM equipped with the beam deflection method is mostly determined by lateral forces. Therefore, on the atomic scale, the lateral forces both in scan direction as well as perpendicular to the scan direction are measured.

Finally, it should be noted that the correct calibration of the bending and the tor-



Figure 3.72: **a)** Principle of the scanning force microscope. Bending and torsion of the cantilever are measured simultaneously by measuring the lateral and vertical deflection of a laser beam while the sample is scanned in the x-y-plane. The laser beam deflection is determined using a 4-quadrant photo diode: (A+B)-(C+D) is a measure for the bending and (A+C)-(B+D) a measure for the torsion of the cantilever, if A, B, C and D are proportional to the intensity of the incident light of the corresponding quadrant. **b)** The torsion of the cantilever (middle) is solely due to lateral forces acting in x-direction, whereas both forces acting normal to the surface (F_z) as well as acting in plane in y-direction (F_y) cause a bending of the cantilever (bottom).

sion of the cantilever is terms of lateral forces is much more difficult than the conversion of the cantilever deflection into the normal force which can simply be performed by applying Hook's law. The procedure which we used for the quantitative analysis of lateral force microscopy experiments is described in Ref. [17].

The Mechanics of a Nanometer-Sized Contact

As we will see later, the exact contact area of tip and sample as a function of the applied loading force has to be known in order to compare the data obtained by scanning force microscopy with theoretical models. Scanning force microscopy, however, does not allow an independent measurement of the contact area. Therefore, we use tips with a geometrically well defined apex and calculate the effective contact area on the basis of continuum elasticity theory. The mathematically simplest case is a spherical tip which is in contact with a flat surface (the so-called *Hertzian contact*, see Fig. 3.73). Without adhesion, the contact area-load dependence $A(F_1)$ for a sphere on a flat surface can be



Figure 3.73: a) Macroscopic tip-sample contact. The contact area-load dependence is difficult to describe due to the irregular shape of both tip and sample surface. b) Geometry of a Hertzian contact (see text). R = radius of the tip apex, a = radius of the contact area.

determined by the Hertzian theory to

$$A = \pi \left(\frac{RF_1}{K}\right)^{2/3},\tag{3.12}$$

where R denotes the radius of the sphere, F_1 the externally applied loading force and

$$K = \frac{4}{3} \left(\frac{1 - \nu_1^2}{E_1} + \frac{1 - \nu_2^2}{E_2} \right)^{-1}$$
(3.13)

the "effective elastic modulus" of the contact (ν_i = Poisson's ratio, E_i = Young's modulus of sphere and flat, respectively; i = 1,2) [11]. A mathematically complete description of the case including adhesion is difficult and does not lead to a simple analytical formula for A. If, however, the radius of the sphere is small (some tens of nanometers), sphere and flat are not too soft and the surface energies of the materials are not too high, as it is often the case in SFM, the additional influence of the attractive forces can be considered by introducing an effective normal force $F_n = F_1 + F_0$. F_0 denotes the sum of all attractive forces and is measured by force-distance curves. Equation (3.12) then changes to

$$A = \pi \left(\frac{R(F_{\rm l} + F_{\rm 0})}{K}\right)^{2/3} = \pi \left(\frac{RF_{\rm n}}{K}\right)^{2/3}.$$
(3.14)

This model will be referred to as the Hertz-plus-offset model in the following.⁹ A detailed discussion of the contact mechanics of a nanometer-sized Hertzian contact can be found in Refs. [11] and [19].

⁹The limit of small tip radius, not too soft materials and not too high surface energies is usually referred to as the *DMT limit*, named after B. V. Derjaguin, V. M. Muller, and Y. P. Toporov who



Figure 3.74: Three examples of specially prepared tips with spherical tip apexes of a) 21 ± 5 nm, b) 35 ± 5 nm, and c) 112 ± 5 nm tip radius.

Equation (3.14) only applies for tips with exactly spherical apex. Our investigations (see the next section) showed that the contact area-load dependence follows only occasionally such a 2/3-power law if SFM tips are used as supplied by the manufacturer. Many different power laws ranging from a power of 1/3 to more than 1 were found. Consequently, tips of exactly defined spherical tip apex and a tip radius which is known with nanometer accuracy are mandatory for quantitatively reproducible friction force measurements.

This was realized by covering doped single-crystalline silicon tips (apex radii 5-15 nm without coating) with a layer of amorphous carbon in a transmission electron microscope. Molecules from the residual gas are ionized in the electron beam and accelerated to the tip end. There, the molecules spread out evenly due to their charge, forming a well-defined spherical tip end. The exact composition of the contamination layer is unknown; measurements by means of energy dispersive x-ray emission, however, suggest that the tip consists mainly of amorphous carbon. With this method, we could successfully produce tips with radii from 7 nm up to 112 nm. Three examples of tips prepared using the method described above are shown in Fig. 3.74; more details on the preparation procedure can be found in Ref. [20].

Typical normal forces F_n occurring in an SFM during measurement are between 1 nN and 100 nN, resulting in contact areas of a few nm². This means that only some tens or some hundreds of atoms are in direct contact. The deformations of tip and sample are then fully elastical, i.e. fully reversible. Therefore, with an SFM, we are

first derived a contact mechanical theory for a sphere on a flat which applies under these conditions [18]. Their original theory, however, cannot be solved analytically, and Eq. (3.14) represents an approximation to their more exact formalism. In order to point out this difference, we named the above introduced approximation the *Hertz-plus-offset model*, which nevertheless only applies in the DMT limit.

able to study wearless friction of a quasi-point contact.

Amontons Laws at the Nanometer Scale

Amontons macroscopic friction laws (i) and (ii) can be condensed in the well-known equation

$$F_{\rm f} = \mu F_{\rm n}.\tag{3.15}$$

Here, $F_{\rm f}$ denotes the observed friction force. μ represents a value which is constant for a given material combination and which is usually referred to as *friction coefficient*. Nevertheless, for the description of the friction behavior of materials at the nanometer scale, it is practical to introduce the mean friction per unit area (the so-called *shear* stress S)

$$S = F_{\rm f}/A. \tag{3.16}$$

Combining Eqs. (3.14) and (3.16), we find

$$F_{\rm f} = \pi S \left(\frac{R}{K}\right)^{2/3} F_{\rm n}^{2/3} = \tilde{C} R^{2/3} F_{\rm n}^{2/3}.$$
(3.17)

In our studies, we investigated different materials such as diamond, amorphous carbon, graphite, C_{60} thin films, mica or germanium sulfide under ambient conditions as well as in dry argon atmosphere (content of water below 1 ppm). Measurements of friction as a function of the normal load showed excellent agreement of experimental data and a theoretical fit according to Eq. (3.17), if S is set constant (cf. Fig. 3.75) [11, 12]. Summarizing the main results, we found that the shear stress S depends only on the used materials and on environmental conditions such as temperature and humidity, but not on the mean contact pressure $p = F_N/A$. These findings result in the following remarkable consequences:

- The independence of the shear stress S from the mean contact pressure $p = F_n/A$ is in flagrant contradiction to Amontons' law (i).
- The observed frictional force F_f is in opposition to law (ii) proportional to the contact area A (cf. Eq. 3.16). In the special case of a Hertzian contact, this leads to a F_f(F_n)-dependence of F_f ~ F_n^{2/3}.
- The continuum elasticity theory applies also at the nanometer scale.

For a classification of the microscopic frictional properties of materials, we introduce the factor $\tilde{C} = \pi S/K^{2/3}$ [cf. Eq. (3.17)] which combines the frictional and elastic properties of tip and sample and which can be regarded as an *effective friction coefficient* for point-contact-like single-asperity friction. The definition of such a coefficient is advantageous for mainly two reasons: (1) If materials with identical intrinsic frictional properties (i.e. the materials show the same friction per unit area), but different



Figure 3.75: The frictional force $F_{\rm f}$ as a function of the normal force $F_{\rm n}$, measured on amorphous carbon with geometrically well-defined spherical tips in argon atmosphere [11]. The data presented in a) were obtained using a tip with a tip radius of $R = 17 \pm 5$ nm, the data in b) using a tip with $R = 58 \pm 10$ nm. Both curves are in excellent agreement with fits according to Eq. (3.17) (solid lines); the dashed lines illustrate the deviation from the linear macroscopic model [Eq. (3.15)]. The offsets of the solid lines from the zero point of the normal force are caused by the experimental uncertainty in the determination of the zero point by means of force-versus-distance curves.

Young moduli are examined in measurements performed with identical tips and loading forces, higher friction will be found on the softer material due to the larger contact area. Materials which have the same \tilde{C} , however, will always show the same friction during an experiment. (2) If the same sample is investigated with tips featuring different apex radii, lower friction will be found in the experiment carried out with the sharper tip. Nevertheless, consideration of the geometry by the calculation of \tilde{C} will lead to identical values of \tilde{C} ; an examle for this will be given in the following.

In the case of the curve presented in Fig. 3.75b), much higher friction is measured compared with Fig. 3.75a); the frictional force at $F_n = 10$ nN, e.g., is about three times larger (~ 15 nN) than the frictional force of only 5.5 nN observed in Fig. 3.75a). However, the calculated numerical value of $\tilde{C} = 0.17 \pm 0.07$ GPa^{1/3} for the second measurement is in excellent agreement with the value of $\tilde{C} = 0.17 \pm 0.09$ GPa^{1/3} derived from the first measurement, demonstrating not only the reproducibility of the method, but also that a knowledge of the geometry of the tip-sample contact is mandatory for the correct quantitative analysis of the frictional behavior of different materials.

Atomic Mechanisms of Friction

In the preceding sections, we learned about the mechanical and tribological properties of nanometer-sized contacts. Questions concerning the fundamental mechanisms of friction at the atomic-scale, however, are still unanswered. Since we are investigat-



Figure 3.76: A simple model for a tip sliding on an atomically flat surface. During scanning the microscope body M is moved in x-direction with the sliding velocity $v_{\rm M}$ for different y-positions.

ing wearless friction on atomically flat surfaces showing a minimum of defects, it is reasonable to expect that the origin for the different friction observed on the different materials can be found in the local arrangement of the atoms at the tip-sample contact area.

If this expectation holds, the friction experienced by the tip should depend, e.g., on the direction with which the tip profiles the sample surface. This *frictional anisotropy* was demonstrated by the nanomechanics subgroup already three years ago on the ferroelectric material triglycine sulfate (TGS) [19,21] and was described in the triannual report 1993-95 (section 3.4.4). In spite of being observed on the micrometer scale, these anisotropy effects strongly suggest that the origin of wearless friction can only be found on the atomic scale. A further instructive example for the influence of the atomic structure on the micrometer scale is guanidinium aluminum sulfate hexahydrate (GASH), which is also a material showing ferroelectric properties. On this material, a conformational (i.e. only geometrical) change within a surface unit cell could be recognized as the reason for the different friction which is observed on the oppositely polarized electrostatic domains (domain contrast) [22].

Considering the atomic structure along the scan direction, the effects mentioned above can be well described with a simple *one-dimensional* spring model [19]. We have seen, however, that on the atomic scale, the SFM is able to detect lateral forces in x-and y-direction independently. Such data can only be explained by extended models of atomic scale friction [23–26].

Figure 3.76 illustrates the model which we applied successfully for the simulation of experimental lateral force maps acquired in atomic-scale SFM experiments. Within this model, the SFM tip is described by a point-like tip. This tip is coupled elastically to the microscope body M with a spring S representing all elastic behavior of the cantilever, the sample and the tip itself. During data acquisition, the body M is scanned with the sliding velocity $v_{\rm M}$ over the sample surface which is simulated by


Figure 3.77: A comparison between experimental and simulated force maps (scan size: 20 Å \times 20 Å). a) and b) show experimental data of the lateral forces F_x and F_y obtained on a freshly cleaved graphite(0001) surface. A simulation of the above force maps using the model described in the text is displayed in c) and d).

the tip-sample interaction potential. A suitable choice of this interaction potential considers that the actual tip-sample contact area is not point-like, but can comprise up to thousand atoms. Finally, energy dissipation which always has to be considered if friction is involved is supposed to be proportional to the *effective* velocity of the tip $v_{\rm tip}$. With these assumptions, the movement of the tip can be calculated by solving numerically the equations of motion, and the lateral forces F_x and F_y acting on the tip can determined [24, 25].

Figure 3.77 contrasts experimentally measured force maps (obtained on graphite) with simulated ones [26]. In spite of the simple model, the simulated force maps [c) and d)] show good agreement with the experimentally obtained data sets [a) and b)]. Even the small features at the left side of the images – effects caused by the tension of the internal springs after changing the scan direction – appear in the simulated force maps.

Please note that all force maps show a structure with a periodicity of 2.46 Å,



Figure 3.78: (a) Gray-scale plot of the used interaction potential representing the hexagonal structure of graphite(0001) (image size: 20 Å \times 20 Å) and three typical calculated paths of the tip in this potential plotted by dots separated by equal time intervals ($\Delta t = 0.1$ ms, $v_{\rm M} = 400$ Å/s). The two dimensional "stick-slip" movement of the tip on the graphite surface is clearly visible: The tip moves discontinuously over the sample surface and stays most of the time in the minima of the interaction potential (dark areas) before it "jumps" nearly instantaneously into the next potential minimum. (b) In this contour plot of the used interaction potential, only the maxima are marked (as open triangles). These locations represent the positions of the carbon atoms which form hexagonal rings with a lattice constant of 1.42 Å. Additionally, all calculated paths of the tip are plotted "time-resolved" as in a) $(512 \times 512 \text{ points})$, demonstrating that the tip stays most of the time at the "hollow" sites" of the surface. Therefore, only the minima of the interaction potential are profiled by the tip. The result is that the measured quantities of atomic-scale SFM experiments – the lateral forces – show a periodic structure with the periodicity of a = 2.46 Å. Consequently, the contact mode imaging process of an SFM results in "hollow-site resolution" instead of "atomic resolution".

whereas the distance of neighboring graphite atoms is only 1.42 Å [26]. This apparent contradiction can be easily explained by the analysis if the movement of the tip on the graphite surface. In Fig. 3.78a, the (calculated) paths of the tip in the interaction potential are plotted for three different scan lines "time-resolved" by points separated by equal time intervals ($\Delta t = 0.1 \text{ ms}$). They reveal that the tip profiles the surface in a kind of zigzag motion; the interruptions in the paths indicate a discontinuous movement ("stick-slip") of the tip in the interaction potential. Only the minima of the assumed tip-sample interaction potential – the "hollow sites" – are in contact with the tip for a significant time, and the maxima – the positions of the carbon atoms – are skipped. If the microscope body M is moved along a specific scan line, the spring S is tightened until the tip "jumps" into the next minimum in order to reduce the spring tension. This "stick-slip" type of movement of the tip is similar to what was found for NaF [14, 24] and MoS₂ [15, 25].

The consequences of this special tip movement on the appearance of the force maps obtained on HOPG can be easily understood considering the "position probability density" [25] displayed in Fig. 3.78b. Obviously, only parts of the "hollow sites" of the hexagonal carbon rings are profiled by the tip, whereas the positions of the carbon atoms cannot be profiled due to the discontinuous tip movement. Therefore, all data acquired in contact mode with force microscopes as described above show only a trigonal structure with 2.46 Å lattice constant which corresponds to the distance and symmetry of neighboring hollow sites. This result can be straightforwardly generalized: "Atomically resolved" SFM micrographs recorded in contact mode consequently always represent "hollow-site resolution" instead of "atomic resolution". For a more detailed discussion of this topic, see Refs. [26] and [27].

From the comparison of the simulations and the experiment, we can further deduce that the decisive springs which determine the nature of the stick-slip movement of the tip are the "interatomic springs" between the atoms close to the tip-sample contact area; the spring constants of the cantilever play only a subordinate role. During tip movement, an elastic dislocation of the atoms at the tip apex and at the sample surface is primarily observed. Thus, a scanning force microscope measures the "true" frictional behavior of a nanocontact.

Effects of the Microscopic Frictional Behavior on the Macroscopic Scale

The aim of the nanomechanics subgroup in 1996-98 was, of course, not only to investigate the microscopic (or nanoscopic, respectively) friction laws, but also to clarify how these microscopic laws are connected with the macroscopic laws from Amontons and Coulomb. The contradiction between Amontons' laws (i) and (ii) and their corresponding counterparts on the nanometer scale can be entirely eliminated if we consider that the effective contact area A_{eff} of two macroscopically flat bodies, which are in relative motion, consists of a huge number of individual nanoscopic contacts. If now the effective contact area of such a macroscopic contact would increase proportionally to the externally applied loading force, the observed frictional force would also increase linearly with the load according to the nanoscopic friction law found above. Such a linear $F_{\rm f}(A_{\rm eff})$ -dependence for macroscopically flat, but microscopically rough surfaces was already demonstrated 1965 by J. A. Greenwood [28]. Interestingly, the apparent contact area of the two bodies does not matter in this case.

Finally, we have to find an explanation for Coulomb's law (iii) on the independence of sliding friction on the sliding velocity $v_{\rm M}$. As already mentioned in the introduction, we could confirm this law also on the nanometer scale within a wide velocity range (see Fig. 3.79). Materials investigated within this study included diamond, graphite and amorphous carbon; the applied velocities ranged from $v_{\rm M} = 20$ nm/s to $v_{\rm M} =$ $24.4 \ \mu {\rm m/s}$. Further details are given in Ref. [13].

Such an independence of the friction on the sliding velocity, however, follows straightforwardly from the atomic-scale model presented in the preceding section. In



Figure 3.79: The measured frictional forces $F_{\rm f}$ as a function of the sliding velocity $v_{\rm M}$ for amorphous carbon, diamond and highly oriented pyrolytic graphite (HOPG). It is clearly visible that $F_{\rm f}$ is independent of the sliding velocity in good approximation. The applied normal force in a) was $F_{\rm n} = 5.2$ nN, the mean observed friction force $\overline{F_{\rm f}} = 6.0$ nN, and a tip with an apex radius of $R = 62 \pm 4$ nm was used. The parameters for the other measurements were: b) $F_{\rm n} = 83.4$ nN, $\overline{F_{\rm f}} = 44.9$ nN, $R = 23 \pm 2$ nm; c) $F_{\rm n} = 3.3$ nN, $\overline{F_{\rm f}} = 12.5$ nN, $R = 23 \pm 2$ nm; d) $F_{\rm n} = 33.0$ nN, $\overline{F_{\rm f}} = 24.4$ nN, $R = 62 \pm 4$ nm; e) $F_{\rm n} = 26.3$ nN, $\overline{F_{\rm f}} = 0.01$ nN, $R = 62 \pm 4$ nm; f) $F_{\rm n} = 8.1$ nN, $\overline{F_{\rm f}} = 0.1$ nN, $R = 35 \pm 3$ nm. The sliding distance was 1 μ m in each direction of the friction loop for all measurements.

this model, we assumed that energy dissipation is proportional to the effective tip velocity $v_{\rm tip}$. Thus, it can be directly concluded from the stick-slip type movement of the tip that the energy is nearly exclusively dissipated during the rapid "slip"-motion [25]. Most of the time, the tip rests nearly static in the potential minimum, and approximately no energy is dissipated. If, however, the "interatomic springs" are sufficiently stretched, a quick "jump" of the tip follows showing a "jump velocity" $v_{\rm jump}$ which depends basically exclusively on the spring constants of the responsible interatomic springs and on the height of the potential barrier between the individual minima. This situation changes only if at very high sliding velocities $v_{\rm M}$ gets comparable with $v_{\rm jump}$. In this case, a dependence of the observed friction on $v_{\rm M}$ would be expected [13]. Experiments in this velocity range, however, could not be realized in a defined manner in SFM experiments up to now for instrumental reasons.

3.4.2 Dynamic Scanning Force Microscopy

W. Allers, H. Hölscher, A. Schwarz, and U. D. Schwarz

In the preceding section, we have seen that static scanning force microscopy (static SFM) operated in the repulsive contact mode in air or vacuum is able to resolve atomic scale structures. Since typical loading forces exerted by the tip on the sample during measurement are of the order of 1-100 nN, up to several hundred tip atoms are in contact with the sample surface (see p. 101). Atomic scale images are therefore interpreted as the periodicity of the ordered surface [29] and usually show only unit cell resolution [30, 31]. Additionally, atomic resolution on highly reactive semiconductor surfaces is particularly difficult to obtain in contact mode [32, 33].

On the other hand, it was experimentally shown that dynamic SFM provides an opportunity to observe single atoms within the surface unit cell as well as point defects. In dynamic ("non-contact") mode SFM, a stiff cantilever, which possesses a sharp tip at its free end, is oscillated with a fixed amplitude at its resonance frequency in ultrahigh vacuum slightly above the sample surface. [34] This resonance frequency is affected by the force field acting between tip and sample; image formation is accomplished by profiling the sample surface with a constant frequency shift Δf of the actual resonance frequency f from the eigenfrequency of the free cantilever f_0 . Atomic resolution with dynamic scanning force microscopy was first achieved in 1995 on Si(111)-(7×7) by Giessibl [35]. Bammerlin *et al.* [36] and Fukui *et al.* [37] were able to image insulators with atomic resolution.

In this section, we will first present selected examples of atomically resolved data which we obtained on the 1×1 -relaxed (110) surface of the III-V-semiconductor InAs, on the graphite(0001) surface, and on the (111) surface of a xenon thin film deposited on graphite(0001). All results were obtained with our newly developed, home-build low temperature ultrahigh vacuum scanning force microscope, which is described in detail in section 3.5.3 as well as in Ref. [38]. Following this experimental section, we will focus on the physics of dynamic SFM, explain its principle mechanism, explore different methods for the calculation of the frequency shift and show how information about the tip-sample interaction potential can be obtained from the analysis of the frequency shift as a function of the mean tip-sample distance.

Atomic Resolution of $InAs(110)-(1 \times 1)$ in Dynamic SFM

A. Schwarz, W. Allers, and U. D. Schwarz

Experimental

The samples were n(S)- or p(Zn)-doped material ($N \approx 3 \times 10^{18} \text{ cm}^{-3}$), cleaved in situ at room temperature and transfered to the microscope which then was lowered into the cryostat. The n-doped (Sb: $8 \times 10^{17} - 5 \times 10^{18} \text{ cm}^{-3}$) silicon tips [39] used for the experiments were cleaned by argon sputtering. Resonance frequencies f_{res} of the cantilevers were on the order of 170 kHz with spring constants k of 35 N/m. The instrument was operated in the dynamic mode, based on the frequency modulation (fm) technique [34], keeping the frequency shift Δf as well as the vibration amplitude A constant. All data are raw data (bandwidth of our fm electronics ≈ 4 kHz), only the lateral distortion due to different piezo sensitivities in x- and y-direction has been corrected.

Simultaneous Resolution of the As and In Sublattice

As most of III-V-semiconductors, InAs crystallizes in the ZnS structure, showing zigzag chains of alternating In and As atoms in the (110) plane (see Fig. 3.80a). At the surface, the As atoms relax outwards and the In atoms inwards, according to the bond rotation model. Therefore, the As sublattice is lifted ≈ 80 pm above the In sublattice



Figure 3.80: a) Top view on the (110) cleavage plane of InAs. As and In atoms alternate in zigzag chains along the $[1\bar{1}0]$ -direction. The lattice constant is a = 606 pm in [001]-direction and b = 428 pm in $[1\bar{1}0]$ -direction. b) Side view of the relaxed surface. The surface is As-terminated with the In atoms lying 80 pm below the As atoms. The dangling bonds, indicated by ellipses, point into the vacuum region. Due to a charge transfer from the In atoms to the As atoms, the As dangling bonds are filled while the In dangling bonds are empty.



Figure 3.81: Dynamic SFM image of InAs(110)-(1×1); the image size is 5.8 nm × 7.1 nm. The bright protrusions are interpreted as As atoms. Parameters: T = 14 K, k = 36 N/m, $f_{\rm res} = 160$ kHz, $A = \pm 12.7$ nm, $\Delta f = -60.5$ Hz.

(Fig. 3.80b). Consequently, the zigzag chains of alternating In and As atoms are not visible in most dynamic SFM images, but rows of bright protrusions which can be interpreted as the As atoms (see Fig. 3.81). The same observation was made by Ohta *et al.* [40] on InP.

Figure 3.82 shows a series of images, where the tip-sample distance was decreased by changing the frequency shift Δf in steps to more negative values from a) to c). At the same time the corrugation amplitude in [001]-direction increases from 9 ± 2 pm (a) to 18 ± 2 pm (c) and in [110]-direction from 4 ± 2 pm (a) to 9 ± 2 pm (c). A dark and strongly localised depression between As atoms along the [110]-direction is visible in c), whereas these depressions (marked by a white circle) are much weaker in b) and disappear in a). All three images were scanned one after another without any tip changes.

We assign the additional contrast of the dark depressions to the In sublattice for three reasons: First, on the assumption that the bright protrusions represent the position of As atoms, the dark depressions occur exactly at one of the two possible positions of the In atoms. Second, the additional contrast becomes visible only at smaller tip sample distances, which is consistent with the fact that the In atoms are located 80 pm below the As atoms due to the relaxation. Finally, the perpendicular distance between the minima of the dark depressions (the potential positions of the In



Figure 3.82: Dynamic SFM of n-InAs(110) of the same surface area (image size: 2.5 nm \times 2.5 nm) at different tip-sample distances. With decreasing tip-sample distance, the corrugation amplitudes increase by a factor of two, and a dark depression between the bright protrusions becomes visible in b) and c), which can be attributed to an additional contrast caused by the In sublattice. The perpendicular distance between the protrusions and depressions in c) can be measured to be \approx 110 pm (distance between the two parallel lines in c). Parameters: T = 78 K, k = 38 N/m, $f_{\rm res} = 176$ kHz, $A = \pm 11.0$ nm, $\Delta f = -361$ Hz (a), -418 Hz (b) and -447 Hz (c).



Figure 3.83: n-InAs(110) imaged with a tip different from the one used in Fig. 3.82. Similar to the situation there, an additional contrast is visible within the unit cell. With the present tip, however, the additional contrast is imaged as a second protrusion with a lower corrugation amplitude, while in Fig. 3.82 the additional contrast is imaged as a depression. Parameters: $T = 14 \text{ K}, k = 36 \text{ N/m}, f_{\text{res}} = 160 \text{ kHz}, A = \pm 12.7 \text{ nm}, \Delta f = -37 \text{ Hz}, \text{ a}$) image size: 2.2 nm \times 3.2 nm, scan angle: 0°, b) image size: 2.5 nm \times 3.2 nm, scan angle: 45°.

atoms) and the maxima of the bright protrusions (the positions of the As atoms) as indicated by two lines in Figure 3.82c), is ≈ 110 pm, which is in good agreement with LEED experiments (100 pm) [41] and theoretical calculations (130 pm) [42].

Not every tip is suitable to image the In sublattice. Often, the tip simply crashes if its distance from the surface is reduced too much. With a different tip we could also observe a second protrusion within the unit cell (see Fig. 3.83). Clearly, the zigzag structure along the $[1\bar{1}0]$ -direction is visible in this image. Because of the different contrast from the In atoms, we conclude that the contrast on InAs(110) not only depends on the tip-sample distance (i.e., on Δf), but also significantly on the tip. Finally, we would like to recall that such a simultaneous observation of both atomic species is different from the usual situation in scanning tunneling microscopy, where measurements on similar III-V-semiconductors (e.g., GaAs [43]) show either the anion (As) sublattice by imaging the filled state or the cation sublattice by imaging the empty states.

What might be the reason for the different kinds of contrast observed? The contrast formation in dynamic scanning force microscopy at the atomic scale is not well understood yet. Perez *et al.* [44] pointed out that the contrast on semiconductors depends on the existence and direction of dangling bonds at the tip apex. In spite of using the same type of silicon tips for all measurements the foremost tip atoms are likely to have a different atomic configuration. A (charged) dopant atom close to the tip might additionally influence the contrast. Moreover, it is important to note that every tip crash can change the atomic configuration at the very end of the tip, e.g., sample material can be picked up and change the contrast drasticially.

A possible mechanism explaining the observed contrasts might be as follows: Since the relaxation is accompanied by a charge transfer between both species at the surface, the As atoms (anions) are negatively and the In atoms (cations) positively charged. Due to a dopant atom or picked up sample material the very end of the tip can also carry a permanent charge, giving an electrostatic contribution to the total (attractive) tip-sample interaction. This electrostatic contribution is reversed above As and In atoms, respectively: A negatively charged tip would lead to a repulsive electrostatic interaction above the negatively charged In atoms. On the other hand, if the tip is easily polarizable or image charges are present, the electrostatic component of the tip sample interaction will be attractive for both species and the As and In atoms are imaged as protrusions. Theoretical studies have to be performed in order to clarify these issues further.

Arsenic Vacancies

In Fig. 3.84, only one type of contrast is visible, similar as to the situation found in Fig. 3.81. However, three protrusions, marked with white circles, are missing within the perfect lattice. If the assumption introduced above was correct that only the As sublattice is imaged, all missing protrusions can be interpreted as As vacancies.



Figure 3.84: Dynamic SFM image of InAs(110); the image size is 10.3 nm × 14.1 nm. As in Fig. 3.81, the white protrusions are interpreted as As atoms. Consequently, the three missing protrusions on n-InAs marked by white circles have to represent As vacancies. Parameters: T = 14 K, k = 36 N/m, $f_{\rm res} = 160$ kHz, $A = \pm 12.7$ nm, $\Delta f = -63.2$ Hz, scan angle: 30°.

Additionally, we note that no distortions are visible around the vacancies. This observation is consistent with theoretical calculations performed by G. Schwarz *et al.* [45] for InP(110), a material which is structurally very similar to InAs. For this material, they found that only the nearest neighbours of a missing phosphorous atom (i.e. the In atoms) relax substantially.

Contrast Inversion

With the image series presented in Fig. 3.85, another point defect is investigated. The tip-sample distance is decreased from a) to c) by reducing the frequency shift Δf . No tip change occurred during scanning. In a), the image with the largest tipsample distance, the resolution is very low and only a slight depression is visible. At smaller separations [images b) and c)], the contrast is much higher. Obviously, the two neighboring As lattice sites which are aligned in $[1\bar{1}0]$ -direction and marked with X (X sites) are affected. In b), the protrusions of the X sites are found to be 6 pm lower than the protrusions of the undisturbed As atoms. Surprisingly, a further reduction



Figure 3.85: a) to c): Image series which shows an inversion of the contrast in the vicinity of a point defect as a function of the tip-sample distance. The point defect shows a mirror symmetry in [001]-direction and is tentatively located at the In site between the As sites marked with X. Possible candidates for point defects exhibiting such a behavior might be, e.g., an In vacancy, an impurity or a substitutional arsenic atom. Parameters: T = 14 K, k = 36 N/m, $f_{\rm res} = 160$ kHz, $A = \pm 12.7$ nm, image size: 3.9 nm × 4.4 nm, $\Delta f = -31.6$ Hz (a), -39.5 Hz (b), and -44.7 Hz (c). (d) This graph illustrates how a contrast inversion would be possible (see text).

of the tip-sample distance (cf. Fig. 3.85c) leads to a contrast inversion compared with the surrounding As atoms: Both "X"-protrusions appear now 10 pm higher than the undisturbed As atoms. Furthermore, the depression marked with S lies in b) 18 pm below the equivalent points of the undisturbed lattice, while this difference is reduced to 8 pm in c). The corrugation amplitude in $[1\bar{1}1]$ -direction increases approximately by a factor of two from b) to c).

At this stage, it is not possible to specify the type of the observed point defect exactly. The point defect could be directly located at the X sites, e. g., an adsorbate or two neighboring substitutional atoms. The latter, however, is very unlikely, while other adsorbates observed on this surface appeared not so localized and could be moved under the influence of the tip. Therefore, it is more likely that the two X sites are As atoms which are influenced by a point defect located at the position of the In site between them. Defects at this location such as an In vacancy, a substitutional atom or an impurity should cause changes in the lattice which exactly show the observed mirror symmetry in [001]-direction.

Figure 3.85d illustrates a possible mechanism for the observed contrast inversion of the X sites compared with the height of the undisturbed As atoms. Above the As atoms, the tip-sample-interaction might be stronger for large separations, but, since the interaction potential is steeper above the X sites, contrast inversion occurs at the point of intersection between the potential above the As atoms and the potential above the X sites. It is important to note that this effect cannot be explained by two potentials with identical shape, which are simply shifted against each other in z-direction.

Although the nature of the observed point defect cannot be determined with absolute certainty, the above results have an important impact on data interpretation in dynamic scanning force microscopy: From the obtained data, it is not possible to decide whether the atoms located at the X sites are situated below (as suggested by Fig. 3.85b) or above (cf. Fig. 3.85c) the As terminated surface. As a consequence, it has to be accepted that constant frequency shift images do not contain purely topographical data on the atomic scale. Site dependent chemical reactivity or elastic properties might play an important role.

Charged Defects on p-InAs

The surface of p-doped InAs(110) investigated at room temperature is shown in Figure 3.86. For this experiment, the tip was not cleaned by argon sputtering prior to measurement, i.e. the tip can be assumed to be covered with a layer of native silicon oxide. Two types of defects are visible in the image; both of them are not localized on one lattice site. The first type of defect is imaged as dark area, the other as bright area. Because of the often radial symmetry and the intact atomic structure, the bright areas are tentatively assigned to the influence of the Coulomb potential of subsurface dopant atoms. As an additional hint which supports such an assumption might serve the observed diameters of the bright areas: They are typically of the order of the screening length which is $\lambda \approx 2$ nm for p-doped InAs with $N_D \approx 3 \times 10^{18}$ cm⁻³.

Within the dark areas, however, the atomic structure seems to be distorted. We preliminarily interpret these areas as one or more charged As vacancies. Such charged As vacancies on p-InAs are already well-known from scanning tunneling microscopy experiments [46]. As in the above case of the dopant atoms, the charged vacancies cause an additional contrast due to their Coulomb potential, but with inverse sign,



Figure 3.86: Dynamic SFM micrograph of a p-doped InAs(110) sample. Beside the atomic corrugation, additional bright and dark contrasts are visible. The bright areas show an intact atomic structure and are interpreted as being caused by the Coulomb potential of subsurface dopant atoms, while the dark areas are attributed to the presence of charged As vacancies. Parameters: T = 300 K, k = 33 N/m, $f_{\rm res} = 149$ kHz, $A = \pm 15.3$ nm, $\Delta f = -355$ Hz, image size: 12 nm × 18 nm.

since Zn dopant atoms are negatively and As vacancies on p-doped material positively charged. Therefore, we can conclude that the electrostatic tip-sample interaction is not due to an induced mirror charge in the tip. A trapped charge or a polarization by the charged defect of the tip would lead to an inverted contrast for different polarities.

Despite the fact that sulfur dopant atoms are charged on n-doped material and As vacancies are also expected to be charged, no contrast from charged defects could be detected on n-InAs. In the case of the dopant atoms, a possible explanation for this fact might be that their Coulomb potentials are smeared out due to the much larger screening length on n-doped material of $\lambda \approx 8$ nm. Since the mean distance between two dopant atoms for a dopant concentration of $N_D \approx 3 \times 10^{18}$ cm⁻³ is 7 nm, their Coulomb potentials strongly overlap. On the other hand, the charge state of the As vacancies is influenced by band bending at the surface which is caused by the contact potential difference between tip and sample. Additionally, we have to consider the possibility that the sensitivity to charges might strongly depend on the tip. Please recall in this context that the image presented in Fig. 3.86 was taken with an unsputtered tip covered with a layer of native oxide, while all the other measurements presented above were acquired with sputtered tips.

Atomic Resolution on van der Waals Surfaces with Dynamic Scanning Force Microscopy: Graphite(0001) and Xenon(111)

W. Allers, A. Schwarz, and U. D. Schwarz

The recent progress in dynamic mode SFM enabled not only, as we have seen in the preceding section, the atomic-scale imaging of semiconductor surfaces, but has also led to a previously unreached resolution on insulating samples such as ionic crystals [36] and oxides [37]. Despite this success, weakly interacting systems remained challenging, and various authors suggested that strong interactions like dangling bonds or repulsive contact forces are essential for observing atoms with dynamic SFM [44,47]. In contrast to this prediction, we have succeeded for the first time to image the atomic structures of weakly interacting van der Waals surfaces, namely highly oriented pyrolytic graphite (HOPG) and xenon(111). Especially the latter material, the xenon, represents an interesting class of materials for fundamental investigations since it is purely bound by van der Waals forces.

The above mentioned assumption that strong interaction forces between tip and sample are required for atomic resolution imaging is supported by the fact that until now atomic resolution in dynamic mode could only be achieved on ionic crystals, oxides and semiconductors which all exhibit strong attractive forces. The extremely stable conditions required for the successful imaging of the (conducting) HOPG and the insulating and weakly bound noble gas crystal were again realized by operating our home-built microscope at low temperatures in ultrahigh vacuum; the instrument is described in detail in Sect. 3.5.3. The rectangular-shaped cantilever used for the acquisition of the data presented in Figs. 3.87 and 3.88 was made of monocrystalline silicon and exhibits a spring constant of 35 N/m and a resonance frequency of 160 kHz. The tip was sputtered *in situ* with argon ions prior to the measurements.

The HOPG crystal was cleaved in the preparation chamber of our UHV system at room temperature (base pressure below 10^{-9} mbar) and was inserted into the cold microscope in the main chamber (base pressure below 10^{-10} mbar) immediately after cleavage. The HOPG crystal was then scanned at a temperature of 22 Kelvin maintaining a constant cantilever resonance amplitude of A = 88 Å (see Fig. 3.87a). The frequency shift Δf from the unperturbed cantilever oscillation $f_0 = 160$ kHz was kept at -63 Hz. The analysis of Δf as a function of the mean tip-sample distance clearly shows that on HOPG, for such values of A and Δf and with this type of tips, the instrument is fully working in the non-contact regime, i.e. that the influence of repulsive interaction is negligible. The low temperatures were vital to achieve the necessary signal to noise ratio: They ensure thermally stable conditions in the microscope and reduce the thermal noise of the cantilever. For different tips and scan angles, we always imaged a structure with sixfold symmetry of the observed maxima and with an apparent corrugation of less than 15 pm. Individual maxima have a distance of 2.46 Å, corresponding to the values found by scanning tunneling microscopy (STM)



Figure 3.87: a) Atomic-scale image of highly oriented pyrolytic graphite (HOPG) obtained with dynamic scanning force microscopy; image size: 53 Å × 53 Å. A structure with sixfold symmetry of the observed maxima and 2.46 Å period was observed with all tips used, corresponding to the symmetry and distances found by STM and contact SFM (cf. Sect. 3.4.1. Image b) displays the surface of a xenon thin film, grown at low temperatures on the HOPG substrate presented in a). The maxima can be identified as the positions of the surface xenon atoms. Parameters: Resonance frequency of the cantilever used in both images: 160 kHz; frequency shift $\Delta f = -63$ Hz (a) and -92 Hz (b); cantilever resonance amplitude A = 88 Å (a) and 94 Å (b), respectively; temperature T = 22 K.

and contact SFM (cf. Sect. 3.4.1). Further details on the graphite measurements can be found in Ref. [48].

For the subsequent study of xenon, xenon atoms were deposited as a thin film on the graphite crystal. Graphite was chosen as substrate since its (0001) plane (which represents a weakly interacting van der Waals surface, as we have seen above) is expected to have only negligible influence on the electronic configuration of the adsorbed xenon atoms. For film deposition, the HOPG substrate was exposed to a 1.3×10^{-5} mbar atmosphere of the noble gas for three minutes at temperatures well below 60 Kelvin. The thickness of the resulting xenon film is not exactly known; force microscopy images acquired on large scales, however, did not show any uncovered areas of the substrate.

Within the film, the atoms are, as in all noble gas crystals, entirely bonded by van der Waals forces, which is why their electronic configuration stays nearly the same as for free atoms. This represents a major difference to other materials, even if compared with materials exhibiting van der Waals surfaces such as, e.g., HOPG, where (0001) oriented layers stick together by van der Waals forces. Inside an individual layer, however, the carbon atoms are firmly bonded by sp²-hybrids and have lost their spherical geometry. As a result, properties of solid xenon such as its electronic structure, bonding strengths, melting and interaction with other materials are comparatively easy to



Figure 3.88: Three-dimensional perspective representation of a 27 Å \times 27 Å large area of the xenon(111) surface, clearly showing the sixfold symmetry of the surface structure. The distance between individual maxima was determined to be 4.5 Å \pm 10 %, the corrugation to \approx 25 pm. Imaging parameters were the same as in Fig. 3.87b.

describe, making it an interesting model system for a direct comparison of quantitative experimental data with results of theoretical considerations including computer simulations.

Figure 3.87b) shows an atomic-scale image of the xenon film; image size (53 Å \times 53 Å) and acquisition temperature (22 K) are the same as for the measurement presented in Fig. 3.87a). Additionally, Fig. 3.88 shows the same surface at a higher magnification (image size: 27 Å \times 27 Å) in a three-dimensional perspective representation, leading to a better visualization of individual surface unit cells. The observed maxima are arranged in a structure with sixfold symmetry, as expected for atoms at the (111) surface of a material crystallizing in fcc structure [49]. The distance between individual maxima was determined to be 4.5 Å \pm 10 %, which is in good agreement with the nearest neighbor distance in a xenon single crystal (4.35 Å) [49]. Therefore, the positions of the maxima can be identified as the positions of the xenon atoms at the surface. The apparent atomic corrugation was about 25 pm, which is considerably

larger than the corresponding value found above for HOPG, presumably due to the large atomic radius of xenon of 2.16 Å [50].

To conclude this section, we have demonstrated that it is possible to image weakly interacting van der Waals surfaces by dynamic SFM with the example of graphite(0001) and xenon(111). Especially the success on the noble gas crystal xenon enables new atomic-scale experiments of considerable theoretical interest on materials previously inaccessible for direct microscopical methods and clarified some of the open questions in the current discussion on the imaging mechanism in dynamic scanning force microscopy.

Analysis of the Frequency Shift in Dynamic Force Microscopy

H. Hölscher and U. D. Schwarz

The Principle of Dynamic Force Microscopy

As already announced in the introduction to this section 3.4.2, we will focus in the following on the analysis of the imaging process and the contrast mechanism. The principle of dynamic force microscopy and the notations used in this article are shown in Fig. 3.89. A cantilever is vibrated with a fixed resonance amplitude A at its resonance frequency f. This resonance frequency is different from the eigenfrequency f_0 of the free cantilever due to the influence of the interaction force between the tip and sample and changes while changing the resonance amplitude A or the distance d between the sample and the tip if the cantilever is undeflected. Since this distance is varied in the experiment by moving the position of the cantilever support relative to the sample surface, d will be called the support-sample distance in the following. Additionally, the tip-sample distance at the point of closest approach is denoted as D.

The reason for the shift of the cantilever resonance frequency can easily be understood by looking at the potentials plotted in Fig. 3.90. If the cantilever is far away from the sample surface, the tip moves in a parabolic potential (dotted line), and its oscillation is harmonic. In such a case, the tip motion is sinusoidal and the resonance frequency is given by the eigenfrequency f_0 of the cantilever. If, however, the support-sample distance d is reduced, the potential which determines the tip oscillation is modified and given by an effective potential (solid line) represented by the sum of the parabolic potential and the tip-sample interaction potential (dashed line). This effective potential differs from the original parabolic potential and shows an asymmetric shape, contrary to the symmetric shape of the parabolic potential. Consequently, the resulting tip oscillation becomes anharmonic, and the resonance frequency of the oscillation now depends on the oscillation amplitude A. Since the effective potential experienced by the tip changes with the support-sample distance d, the frequency shift Δf depends on both parameters.

Calculation of the Frequency Shift

An exact way to calculate the frequency shift is to solve the equation of motion of the tip in its potential and to determine the deviation of the actual resonance frequency from the eigenfrequency of the cantilever. This equation of motion is given by

$$m_z \ddot{z} + c_z (z - d) = F_{\text{int}}(z),$$
 (3.18)

where $m_z = c_z/(2\pi f_0)^2$ is the effective mass and c_z the spring constant of the cantilever; the term $F_{\text{int}}(z) = -\frac{\partial V_{\text{int}}(z)}{\partial z}$ describes the interaction force between tip and sample.

A widespread method to handle a nonlinear equation of motion like Eq. (3.18) is to simplify the nonlinear term F_{int} and to solve the resulting equation of motion analytically. In the case of dynamic force microscopy, the interaction force F_{int} is usually expanded into a Taylor series. If adequate oscillation amplitudes are applied, $F_{\text{int}}(z)$ can be approximated within the range of the oscillation $(d_0 - A \le z \le d_0 + A)$ by the first two terms of the Taylor series. However, for general interaction forces $F_{\text{int}}(z)$, such oscillation amplitudes are normally very small. Calculating the solution of the simplified differential equation, it can be shown that the frequency shift is approximately given by

$$\Delta f \approx -\frac{f_0}{2c_z} \frac{\partial F_{\text{int}}(d)}{\partial z}.$$
(3.19)

This equation is the origin of the often repeated statement that the frequency shift occurring in dynamic force microscopy is a measure for the force gradient of the tip-sample force. It has, however, to be emphasized that according to the derivation of Eq. (3.19), this is for general tip-sample forces only true in the *limited case* of *small* resonance amplitudes.



Figure 3.89: Scheme which illustrates the notations used in this article. The cantilever oscillates with a resonance amplitude A. During one cycle, the tip approaches the sample to the nearest tipsample distance D. The distance between the tip and the sample at the point where the cantilever in undeflected is called the support-sample distance d.



Figure 3.91: Comparison between the different methods to calculate the resonance frequency shift for the oscillation amplitudes 10 Å (a) and 100 Å (b) at the example of the van der Waals force Eq. (3.20). The exact result obtained numerically from Eq. (3.18) is marked by black circles in (a) and by squares in (b). The calculation using the force gradient method Eq. (3.19) (dashed lines) is only useful in the limit of small amplitudes and large tip-sample distances (see text).

In order to show the quality and the region of validity of the often used approximation (3.19) the result of this formula is compared with the "exact" result obtained by the numerical solution of the equation of motion Eq. (3.18). Both methods are applied to a long-range tip-sample force describing the *van der Waals* force between a tip with radius R and a flat sample surface. This force is often described by [50]

$$F_{\rm vdW}(z) = -\frac{A_H R}{6z^2},$$
(3.20)

where A_H represents the Hamaker constant. Typical parameters are R = 120 Å and $A_H = 0.1$ aJ. Retardation effects which would affect the interaction at distances larger than about 100 Å in a real physical system are neglected for simplicity.

The result demonstrates that the interpretation of the frequency shift as a force gradient is only satisfactory for distances greater than 30 Å for a small amplitude of 10

Å (dashed-dotted line in Fig. 3.91a). If the amplitude is increased to 100 Å, the force gradient method fails for all distances up to 30 nm (Fig. 3.91b). This result agrees with the derivation of the frequency shift formula Eq. (3.19). If the variation of the tip-sample force is too large within one oscillation period of the tip, the approximation of $F_{int}(z)$ by the first two terms of the Taylor expansion is inappropriate, and Eq. (3.19) consequently fails. A more detailed discussion can be found in Ref. [51].

Frequency Shift versus Distance Curves

In order to study the influence of the resonance amplitude and the support-sample distance on the frequency shift, we examined the dependence of both parameters on Δf by a comparison of experiment and theory for a graphite sample. The frequency shift was measured as a function of the distance d for different resonance amplitudes and the obtained frequency shift versus distance curves are analysed by simulated data.

The experiments were carried out with our home-built scanning force microscope for operation in ultrahigh vacuum and at low temperatures (see Sect. 3.5.3). The rectangular shaped cantilever used for this study was made of monocrystalline silicon with a spring constant of 38 N/m and an eigenfrequency of 171 kHz. The tip was



Figure 3.92: Comparison of the measured and simulated frequency shift versus distance curves. The frequency shift obtained with a silicon tip and a graphite(0001) sample for different resonance amplitudes is displayed by the symbols (experimental data by A. Schwarz). The simulated curves are calculated according to the procedure described in the text and plotted by solid lines.

sputtered in situ with Argon prior to the measurements. The graphite sample was cleaved in situ at room temperature at a pressure below 10^{-9} mbar and immediately inserted into the cooled microscope. The sample was cooled down to T = 80 K and series of experiments were carried out measuring the frequency shift versus the support-sample distance in dependence of the resonance amplitude. The experimental data presented here was recorded with a zero bias voltage between tip and sample. The obtained frequency shift versus distance curves are presented in Fig. 3.92 for resonance amplitudes between 54 Å and 180 Å. All curves show the same typical behavior. If the cantilever approaches the sample surface, the frequency shift decreases and reaches a minimum. With a further reduction of the support-sample distance the frequency shift increases again and shifts to positive values.

To obtain more information about the tip-sample interaction from the frequency shift versus distance curves it is useful to calculate the frequency shift for a suitable tip-sample interaction force. This can be done by solving the equation of motion of the tip Eq. (3.18). However, to reproduce the experimental data, an appropriate tipsample force F_{int} had to be chosen. The theoretical Δf curves plotted in Fig. 3.92 are calculated using the following tip-sample interaction force.

If tip and sample are far away from each other, the macroscopic properties of tip and sample are most prominent and the tip-sample force is given by long range van der Waals forces which are considered by the force law (3.20). If tip and sample come closer together, the interatomic forces between the foremost atoms of the tip apex and the sample surface atoms become more important. This short range force between the foremost atoms of the tip and nearest sample surface atoms can be recognized by the force of a Lenard-Jones potential

$$F_{LJ} = \frac{12E_0}{r_0} \left(\left(\frac{r_0}{z}\right)^{13} - \left(\frac{r_0}{z}\right)^7 \right)$$
(3.21)

where E_0 is the binding energy and r_0 the equilibrium distance of the Lenard-Jones potential [47,52]. Consequently, the effective non-contact force between tip and sample is often described by the sum of these two forces and given by $F_{NC} = F_{vdW} + F_{LJ}$.

With further approach of the tip to the sample the two bodies come in contact and the elasticity of both bodies has to be taken into account. For small tip radii, this can be done using the *Hertz-plus-offset* model (cf. page 100). The contact force between two arbitrary bodies is given by $F \propto h^{3/2}$, where h is the indentation of the two bodies [53]. The adhesion force at the nanometer scale can be recognized by a small offset. Therefore, the elastic contact force between tip and sample can be described by a force

$$F_c = g_0 (z_0 - z)^{3/2} + F_{ad} \quad \text{for} \quad z < z_0$$
(3.22)

where g_0 is a constant, F_{ad} the adhesion force between tip and sample surface and z_0 the point of contact.

To put the non-contact and contact force together the contact point z_0 is defined to be in the minimum of the non-contact force F_{NC} and the adhesion force is given by the non-contact force in the contact point $F_{ad} := F_{NC}(z_0)$. The calculated frequency shift versus distance curves obtained with this tip-sample force are shown in Fig. 3.21. The used parameters are $A_H = 0.2$ aJ, R = 120 Å, $E_0 = 3$ eV, $r_0 = 3.4$ Å, and $g_0 = 5 \cdot 10^{-5}$ N/m^{3/2}. The good agreement between experiment and simulation demonstrates that the frequency shift can be calculated using Eq. (3.18) and that the introduced force law is suitable to describe the tip-sample force.

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3.5 Instrumental Developments

3.5.1 An UHV-STM for In-Situ Studies of Thin Film Growth

Ch. Witt and M. Bode

Introduction

Scanning tunneling microscopy (STM) is a useful tool to investigate the growth of thin films since it imparts a real-space picture of the morphology after deposition of material onto the surface of the substrate. This can be improved by sequences of images because time-resolved imaging helps the understanding of growth kinetics. Indeed, in order to get dynamical information about particular surface features the evolution of a definite location is of primary interest. For instance, that can be a defect or a nucleation site. The sequential observation of a specific position on the sample surface allows explicit conclusions about nucleation and growth.

The *in situ* preparation of thin films encounters an essential problem: when operating an evaporation source close to an STM, the instrument itself acts as an obstacle. One has to evaporate under an acute angle with respect to the sample surface in order to deposit film material under the tunneling tip [1]. Using this procedure, one cannot exclude coating of the tip which is unfavorable for applications like scanning tunneling spectroscopy (STS) and for stable imaging conditions. Furthermore, an acute beam angle can produce artificial shadows behind steps on a crystal surface. The role of the deposition angle on the growth behaviour is in many cases experimentally unknown. To avoid artifacts, the more common arrangement is desirable where substrate surface and evaporation beam are normal to one another. Another possible way is to remove the sample completely out of the apparatus and to do the preparation in another stage. However, such a mechanical treatment takes a prolonged time for manipulation [2]. Thus, the instrumental issue is to regain locations microscopically on length-scales of nm and to handle the samples macroscopically on scales of at least mm. In this paper we present a new STM design which permits a perpendicular evaporation arrangement at a sufficiently large tip-sample distance without the need to take the sample out of the STM. It allows a sample realignment with a spatial reproducibility of better than 100 nm.

STM-Design

The microscope's basic concept is to retract the sample under an angle of 30° from the scanner. Figure 3.93 shows the instrument's principal geometry. It consists of two parts. The upper part—the microscope's head—contains the scanner and the *in vacuo* preamplifier. The motor for coarse–positioning of the sample is installed in the microscope's body which has outer dimensions of 90 mm × 50 mm × 54 mm. All parts of the instrument are manufactured of the machinable glass ceramic Macor. The materials employed were able to compensate effectively for thermal drift movements. Position 1 is the scanning position. If the sample is retracted from the tip the slope of the coarse positioning motor results not only in a vertical movement but also in a



Figure 3.93: (a) Schematic drawing of the UHV–STM. The microscope is constructed of two parts, the head and the body. The separate head concept enables an easy and quick change of the scanner for different purposes. All parts belonging to the scanning mechanism fit into the head, whereas the ground–body contains the stepper motor for sample positioning. Position 1 and 2 refer to the two treatments the sample undergoes, namely scanning (1) and film-depositing (2). (b) The cut A-B reveals the detailed design of the linear stepper motor. The displacement between positions 1 and 2 is 23 mm.

lateral component. The lateral component of movement is used to place the sample out of the scanning position in the microscope's head down to the preparation position (Pos. 2). The distance between Pos. 1 and Pos. 2 amounts to approximately 20 mm. The evaporation beam passes now vertically through a gap in the scan-head onto the sample surface. The gap has an area of $10 \times 8 \text{ mm}^2$. Then, the specimen will be returned to the original scanning position.

A stepper motor [3] is installed in the microscope's body and includes a sapphire prism with a length of 50 mm as the moving part. Since vibrations of this sapphire piece may result in reduced stability we have chosen a prism with a relatively large edge-length of 15 mm. Its calculated resonance frequency is 17 kHz. A Cu-Be spring as a sample holder is mounted on top of the prism-shaped sapphire. The sapphire is moved by six shear-mode piezo stacks. To provide a well-defined mechanical connection to the sapphire piece, polished Al₂O₃ plates with an area of $5 \times 5 \text{ mm}^2$ are glued on the piezo stacks and the prism faces are optically polished. Two shear piezo stacks are glued on the faces of a V-shaped groove in two planes marked by cut A-B in Fig. 3.93(a), respectively. These piezo stacks are only visible in the cut A-B which is shown in Fig. 3.93(b). After inserting the sapphire prism in this groove two further shear piezo stacks which are glued on a separate Macor beam are pressed on the prism top face. A uniform pressure distribution on each of the six piezo stacks is necessary for reliable operation in UHV and at low temperatures. This is guaranteed by pressing down the Macor beam in its center position, i.e. at equal spacing towards both piezo stacks, with a leafspring via a ruby ball with a diameter of 3 mm. Thus the beam balances the spring force between the two planes A–B. The pressure, and thus the stepper motors' walking velocity can be adjusted by changing the tension of the leafspring. Properly adjusted and triggered with a 1000 Hz pulse (amplitude 400 V), the motor achieves a single step size of ≈ 350 nm and thus an average speed of ≈ 20 mm/min. This results in a total transfer-time, i.e. from Pos. 1 to Pos. 2 and vice versa, of less than 2 min which is fairly reasonable for experiments with surface sensitive samples in UHV.

The scan-head has a height of 30 mm and a contact area with the microscope's body of $50 \times 46 \text{ mm}^2$. It contains a scanning piezo tube coupled to a low noise preamplifier. This keeps the distance between the tunneling tip and the first amplification stage small. The operational amplifier is used in the I-U converter mode with an amplification of 10^8 . The scanner itself, made out of a 1/4"-diameter piezo tube, is designed with a maximum scan range of 3.6μ m with an applied control voltage of 280 V. The complete instrument is covered by a steel cap which was designed closely around the STM's body to provide additional electromagnetic screening and mechanical protection. The only openings of this cap are for sample exchange and tip observation. The lowest mechanical resonance frequency of the piezo tube scanner was measured to be 9.2 kHz. For vibrational isolation the microscope is fixed with six screws on top of a fourfold steel plate-viton stack as a damping stage, as illustrated in Fig. 3.94. Furthermore, the UHV-system is supported by home-built air pressure

damping feet with a resonance frequency of approximately 3 Hz. All electrical connections pass through holes in the steel plates down to a DN100CF-flange on which the whole microscope is mounted. All experiments were performed under UHV-conditions with a base pressure $p < 1 \times 10^{-10}$ torr. The tube-scanner was calibrated by imaging



Figure 3.94: Photograph of the UHV-STM. The stainless-steel plates are separated by Viton rubber rings. In the center part the sapphire piece with a hole for wires can be seen. A DN100CF flange carries the whole setup.

monoatomic steps on Au(111)–films for the vertical sensitivity and by atomically resolving the graphite (HOPG) lattice for the lateral sensitivity. The system investigated was the heteroepitaxial growth of Mn on W(110).

Results

Figure 3.95 shows four STM-images selected out of a series of 18 images in total visualizing the growth of Mn on one particular —nonetheless representative— location of the W(110)-crystal. The clean W(110) substrate exhibits two monoatomic steps roughly along the [001]-direction separating three atomically flat terraces from each other. The steps of 2.35 Å height can also be recognized in the line-section drawn



Figure 3.95: Topography of the clean W(110) substrate and after deposition of 0.6 ML, 1.3 ML, and 2.3 ML Mn. The measurement parameters were U = +0.2 V and I = 0.2 nA.

along the $[1\overline{1}0]$ -direction as indicated by a white line (cf. Fig. 3.95) and plotted in Fig. 3.96. After deposition of 0.6 ML Mn numerous single monolayer high islands can be recognized on the terraces. These islands are elongated along the $[1\overline{1}0]$ -direction of the substrate with an approximate $[1\overline{1}0]/[001]$ length ratio of 3, exhibit a typical length of 100 nm and an apparent height of 2.50 ± 0.12 Å. The rim of these first Mn– layer islands shows some dendritic tendencies. On approximately 60% of these islands the nucleation of second Mn-layer islands with an apparent height of 2.36 ± 0.11 Å can be observed. The latter one exhibit a $[1\overline{1}0]/[001]$ length ratio of 6 and have a smooth rim. The substrate steps are decorated by Mn fingers directed perpendicularly to the steps, i.e along the $[1\overline{1}0]$ -direction. Their lateral dimensions are a factor of 3–4 smaller than those of free–standing first ML islands on the terraces. We found no hint of any reconstruction or dislocation on the Mn islands in the STM-topographs and the belonging line-sections. The absence of superstructures and satellite spots in the LEED-pattern (not shown here) confirms that Mn grows pseudomorphically at this coverage on W(110), i.e. Mn mimics the bcc-symmetry as well as the lattice constant of the underlying substrate in spite of the enormous expansive strain. An exact determination of the misfit between δ -Mn and W is not possible, since δ -Mn



Figure 3.96: Sequence of line-sections (black line) drawn along the same location of the sample (indicated by a line in Fig. 3.95) with increasing Mn coverage. The tungsten substrate and the Mn layers have been differently grey shaded.

is stable only at elevated temperatures and available values for the cubic lattice constant $a_{\rm cub}$ were determined by methods the accuracy of which is not known very well, namely theoretically [6] ($a_{\rm cub} = 2.78$ Å), by extrapolation of high-temperature lattice constants down to room-temperature [5] ($a_{\rm cub} = 2.98$ Å), or by extrapolating data obtained from alloys to zero non-Mn content. In fact, available values of the misfit $f \stackrel{\text{def}}{=} (a_{\rm W} - a_{\rm Mn})/a_{\rm Mn}$ range from f = 6% up to f = 13.6%.

In the STM-image measured at a total coverage of 1.3 ML (Fig. 3.95) several clusters appear on top of second monolayer islands. Their apparent height is 5.8 ± 0.2 Å. The image obtained at 2.3 ML coverage reveals that these clusters serve as nuclei for the growth of roughly rectangular shaped islands. Their step edges are preferentially orientated along the $[1\bar{1}0]$ - and [001]-direction. Furthermore, around the bottom of these Mn clusters or islands the growth of the third Mn monolayer can be observed as examplarily indicated by two circles in Fig. 3.95. Comparison with the line-sections in Fig. 3.96 for coverages of 1.3 ML and 2.3 ML show an apparent height of 2.16 ± 0.15 Å for the third atomic layer and of 4.3 ± 0.1 Å of the rectangular islands. Obviously, the latter exhibits the double apparent height than the former. Therefore, we conclude that the minimal local coverage within the rectangular islands is five atomic layers. We never found any location on the sample which could be attributed to a local coverage of four atomic layers.

In summary, we have designed and tested an STM with the capability of regaining accurately a specific spot of the surface. One of the most promising applications — the in-situ study of thin film growth — has been demonstrated. Such experiments may lead to a better understanding of growth-kinetics of thin films.

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3.5.2 Low-Temperature UHV-STM in High Magnetic Fields

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To increase the spatial as well as the spectroscopic resolution of the STM a large number of developments in scanning tunneling microscopy were performed [1-4]. Much effort is made to enable operation in certain environments such as low temperatures [2,5-10], which additionally leads to a higher stability and a reduced noise level. Many physical properties have to be investigated at low temperatures, often in combination with a magnetic field and UHV (especially quantum effects like standing electron waves [11] or Coulomb blockade [12, 13]).

Our instrument consists of a mechanically very stable home-built microscope with a topographic noise level of 3–5 pm and a spectroscopical resolution of $\leq 7 \text{ mV}$. It is operated in UHV and at temperatures down to 6.2 K; at low temperatures, the thermal drift is lower than 3 Å per hour. The helium bath cryostat is equipped with a pair of superconducting magnets that generate fields of up to 6.5 T perpendicular to the sample surface, or a rotatable magnetic field of up to 1 T. Both tip and sample are easily exchanged under UHV conditions.

UHV System

Figure 3.97 shows a schematic drawing of the whole instrument. In order to achieve a high stability of the tunneling junction a rigid construction of the STM as well as a good external damping is required.

The laboratory room has its own foundation which decouples it from vibrations of the building. This foundation has a large pit for the damping table which supports the cryostat. The three UHV chambers are mounted on the table above the cryostat. The cryostat itself is embedded in sand so that it has no rigid contact with the damping table. The decoupling from the upper UHV part is realized by a flexible edge welded bellow.

The UHV chambers on the sides are equipped for sample preparation and

LEED/Auger analysis. The center chamber is used to transfer samples into and out of the microscope. The pumping system consists of ion getter pumps and Ti sublimation pumps. An additional turbo pump is used during sample preparation when a high gas load is required. The base pressure is below 1×10^{-10} mbar in the whole system and even lower in the UHV insert of the helium-cooled cryostat.

The microscope is transferred between the transfer level and the cryostat insert by a large vertical manipulator. The manipulator consists of an edge-welded bellow with an inner diameter of 98 mm and a total travel of 1400 mm; the microscope is supported from the top flange by a long, thin-walled stainless steel rod. This transfer of the entire STM makes tip and sample easily accessible for transfer. The difficulties involved in this setup, namely the transport of weak electrical signals over long cables



Figure 3.97: Side view of the low temperature ultrahigh vacuum STM/STS system with rotatable magnetic field.



Figure 3.98: Top view of the system.

and the thermal contact between the conical surfaces, have been overcome by now.

Figure 3.98 shows the three UHV chambers from the top view. The sample transfer and handling is realized by horizontal magnetic linear and rotary drives and two wobble



Figure 3.99: Thermal coupling and arrangement of magnets.

sticks.

The left-hand chamber contains a home-built electron beam heater which allows to heat samples to temperatures between 800 and 2500 °C. To transfer tips and samples into and out of the UHV system, a load lock including a cleavage stage is attached to this chamber as well. A commercial UHV evaporator in the right-hand chamber is used to deposit thin films. In order to heat samples during an evaporation, a self-designed resistive heater (up to 550 °C) is fitted to a horizontal magnetic linear motion and rotary drive. The design of the resistive heater allows to heat the samples both during evaporation and LEED/Auger analysis.

Cryostat and Superconducting Magnets

In order to eliminate vibrations from boiling nitrogen, a superinsulated liquid helium (LHe) bath cryostat without external liquid nitrogen shield is used. The cryostat has a usable LHe volume of about 60ℓ and an operation time of up to three days between LHe fillings.

The thermal coupling between the LHe dewar and the cooling stage of the UHV insert is done by He gas in the volume between the UHV insert and the cryostat bore (Figure 3.99). The microscope is mounted to a copper cone that snugly fits a copper countercone in the UHV insert. To increase the thermal contact between the

surfaces, a silver coating was applied to the inner cone at a later stage. This reduced the minimum temperature of the microscope from approx. 8 K to the present value of 6.2 K.

Temperatures are measured at two different positions, on the top of the copper cone and at the bottom of the microscope. Further, we plan to mount a GaAlAs sensor directly beneath the sample. This sensor in a combination with a small resistive heater at the sample stage should allow controlled heating of the sample.

To change the sample or the tip, the microscope is lifted up to the center of the transfer chamber. The exchange is done with a wobble stick; during these few minutes the microscope temperature usually does not exceed 45 K. After the microscope is brought back down into the cryostat it takes between three and four hours until the base temperature of 6.2 K is reached again.

In order to apply a magnetic field rotatable in one plane perpendicular to the sample surface the magnet system consists of an arrangement of two superconducting magnets with fields perpendicular to each other. The vertical field (max. 6.5 T) is generated by a solenoid, the horizontal component (max. 2 T) by a Helmholtz pair. Because of the magnetic interaction of the arrangement, the rotatable field is limited to a maximum of 1 T.

The UHV insert fitted to the bore of the magnets has an inner diameter of 50 mm. To achieve good UHV conditions it is bakeable up to 120 °C with a built-in heater. During bakeout the exchange gas space is evacuated to effectively insulate the UHV insert from the cryostat. The cryostat vessel is filled with liquid nitrogen to prevent the magnets from overheating during bake-out.

STM and Electronics

In order to build a reliably working STM which operates in UHV, at low temperatures, and in high magnetic fields one has to use specially chosen materials. They must be nonmagnetic and critical parts must be carefully chosen with respect to their thermal expansion coefficients closely matching each other.

The microscope body with a length of 84 mm and 32 mm in diameter was made from Macor ceramic. A socket for electrical connections made from copper is mounted on the top of the microscope. It also acts as a thermal and a mechanical connection between the microscope body and the copper cone and allows simple disconnection from the manipulator.

The coarse approach shown in Figure 3.100 is realized by a stepper motor which has already been described in detail elsewhere [14]. The motor consists of six shear piezo stacks which hold a sapphire prism. The sample holder made from copper is mounted on the top of the sapphire prism. To move the prism, one piezo stack after another is sheared by applying a high voltage while the five other stacks hold the prism in position. After all stacks have been moved, they are brought back to their initial position simultaneously, taking the prism with them. The step size varies between two hundred and less than one hundred nanometers, depending on the operating temperature.



Figure 3.100: Design principles of the approach mechanism and the microscope. **a**) side view, **b**) top view. A detailed description was given elsewhere [14].

The scanner which holds the tip consists of a quarter-inch piezo tube with quartered electrodes and has a scan range of about $1.4 \,\mu\text{m}$ at $6.2 \,\text{K}$ with a maximum applied voltage of $\Delta U = 400 \,\text{V}$. The tip itself is mounted to a wedge-shaped carrier that fits into a counterpiece on the scanner. This tip carrier can be inserted and removed using a wobblestick, allowing for simple tip exchange under UHV conditions.

The I-V converter is mounted on the top UHV flange outside the cryostat. The cable length of approx. 2 m between STM and converter necessitates good electrical shielding. Additionally, the conductive heat transfer into the microscope by the wiring has to be minimized. Therefore, custom-made capton insulated shielded stainless steel cables are used for the electrical connections to the microscope. The cables are thermally anchored on the top of the copper cone.

To control the tip position above the sample, a home-built analog feedback partially controlled by the computer is used. The separation of the feedback from the rest of the electronics has the advantage of minimizing noise and of avoiding a damage to the tip in case of computer failure because the feedback loop stays unaffected. The z signal of the feedback and the signals for lateral tip positioning are amplified by commercially available STM electronics [15].

Data acquisition is done by a computer which monitors and controls the actual STM measurement as well as most of the additional parameters of the experimental setup. A new STM software was developed which is able to control the analog feedback loop


Figure 3.101: **a)** Topographic image (raw data) of p-InAs(110) at 9 K (occupied states), size $5 \times 5 \text{ nm}^2$, U = 500 mV, I = 150 pA. **b)** Section along the line indicated in a). The corrugation parallel to the atomic rows is 30 pm. The noise level is found to be 3–5 pm.

and additionally features several scanning and spectroscopic modes. Special modes of operation like distance dependent I(V) spectroscopy for the examination of band gap structures [16] are implemented into the software.

The temperature controller, the magnetic field power supply, the helium level meter, and the lock-in amplifier are connected via an IEEE port to the computer and are simultaneously controlled by the software. This makes it possible to automatically run long, unattended series of measurements with a pre-programmed sequence of external parameters.

For sample preparation the e-beam heater is connected by a self-developed DAC/ADC board to the computer which enables controlled long-term preparation cycles with different temperatures and durations. Heating a single crystal in many small steps results in a very good pressure during preparation because of the low temperatures of the surrounding material. This was helpful in preparing a tungsten single crystal which was used as a substrate for thin film deposition.

First Experimental Results

To show the performance of the new system, topographic and spectroscopic measurements on *p*-type InAs with $N_a(Zn) = 3 \times 10^{18} \text{ cm}^{-3}$ are presented.

The InAs(110) single crystal surface was cleaved in-situ. Figure 3.101 a) shows raw data of a topographic image (occupied states) taken at T = 9 K, U = 500 mV, I = 150 pA. Similar to other III-V semiconductors only one chemical element is seen at a particular bias polarity [17]. The distance between the rows is 6.04 Å while the



Figure 3.102: Spectroscopic data taken on p-InAs(110) at 9 K and in a magnetic field of 3 T perpendicular to the sample surface. **a)** I(V) curve, **b)** dI/dV curve measured simultaneously with a lock-in amplifier. The band gap is found to be 410 ± 20 meV while the Fermi level is located 30 meV beneath the onset of the valence band.

distance of atoms within the rows is 4.27 Å. The corrugation parallel to the atomic rows is 0.3 Å. The noise level as shown in Figure 3.101 b) determined from section lines is found to be 0.03-0.05 Å.

Figure 3.102 shows spectroscopic data taken on p-type InAs(110) with an etched tungsten tip at 9 K in a magnetic field of 3 T applied perpendicular to the sample surface. Voltage and the current of -200 mV and 350 pA were used respectively. The spectroscopic curves were obtained by averaging 16 single data sets on a defect-free area of the surface. In Figure 3.102 a) the I(V) curve is presented while Figure 3.102 b) shows the dI/dV curve which was simultaneously recorded with lock-in technique. The band gap of $410\pm20 \text{ meV}$ was measured which is in good agreement with Varshni [18] and Fang *et al.* [19]. The Fermi level is located 30 meV beneath the onset of the valence band as expected for p-doped material. Measurement of both curves was not affected by applying the magnetic field. Tip-induced band bending [20] does not seem to play a major role in the determination of the spectral width and position of the measured InAs band gap.

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3.5.3 An Ultrahigh Vacuum Scanning Force Microscope for Atomic Scale Experiments at Low Temperatures

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We developed a scanning force microscope (SFM) for operation at low temperatures (LT) in an ultrahigh vacuum (UHV) environment [1]. There are several reasons why it is interesting to operate scanning force microscopes at low temperatures: The thermal drift of the microscope, the creep of the piezoelectric elements used for scanning as well as the thermal oscillations of the cantilever are considerably reduced, thus leading to more stable imaging conditions. The stability and rigidity of the atoms at the sample surface (including adsorbates) as well as at the tip apex is enhanced. Moreover, phase transitions and other low-temperature effects can be studied. Up to now, developments of UHV compatible microscopes have led to several commercially available instruments, but the complexity of the SFM has made cryogenic temperature designs a difficult task: Variation of the temperature and the boiling of the cryogen interferes strongly with the stability and correct alignment of critical parts of the SFM, thus making it difficult to construct a reliably working LT-UHV-SFM, especially if atomic resolution is required. Our microscope features an all-fiber interferometer detection mechanism and achieves atomic resolution in static contact as well as in dynamic non-contact mode from room temperature down to 10 K. Furthermore, the distance-dependent tip-sample interactions can be measured locally in both modes.

Under stable low-temperature conditions, our instrument enables new and unique experiments with atomic resolution in real space on conducting as well as on nonconducting surfaces [2,3]. Selected results are presented in 3.4.2 (atomic structure and defects on n-InAs in the dynamic mode), 3.4.2 (atomic resolution in the static and dynamic mode on a van der Waals surface) 3.4.2 (atomic resolution an a condensed van der Waals crystal) and 3.4.2 (measurement of the tip sample interaction potential). In this section, however, we will describe the design of the complete system, and give only some first results in order to demonstrate the performance of the instrument.

UHV System

When designing the UHV system, good vibration isolation has to be provided. Figures 3.103 and 3.104 show schematic views of the UHV system. The complete UHV and LT equipment is mounted on a table carried by pneumatic damping legs which in turn stand on a separate foundation for decoupling from building vibrations. The bath cryostat is located directly below the UHV chambers, separated from these by a bellow which suppresses the propagation of vibrations from the UHV chambers to the LT part and vice versa. The dewar is installed in a hole in the middle of the table, reaching into a 1.5 m deep pit in the floor (see Fig. 3.103). It is surrounded by sand for isolation from acoustics.



Figure 3.103: UHV- and LT-system, front view.

The UHV system itself consists of three parts. The *preparation chamber* (right chamber in Fig. 3.104) contains a sputter gun and allows sample cleaving. It also serves as a transfer chamber for introducing samples and cantilevers via air lock into the rest of the vacuum system. The design leaves enough space for upgrading, e.g., by evaporation cells. The *analysis chamber* (left in Fig. 3.104) is designed for LEED and Auger measurements. The *main chamber* finally houses the SFM and a vertical transfer mechanism described in the next section.

The vacuum is generated by three ion getter pumps (one for each chamber), two turbo pumps (one for the analysis chamber and the main chamber and the other one for the preparation chamber and the air lock) and one titanium sublimation pump (installed in the main chamber). With this arrangement, pressures better than 1×10^{-10} mbar are currently reached without cooling of the cryostat.



Figure 3.104: UHV- and LT-system, top view.

Cryostat and Vertical Transfer Mechanism

For the LT measurements, the complete SFM is transferred from the main chamber (where tip and sample exchange are carried out *in situ* at room temperature) into the bath cryostat. This procedure has several significant advantages especially for high-resolution atomic-scale imaging. First of all, the cooling of the complete microscope eliminates most of the thermal drift problems which cause misalignments and instabilities since all parts are kept at equilibrium temperature. Low temperatures also reduce the nonlinear temperature dependent creep effects of the piezo actuators. Finally, using a bath cryostat instead of a flow cryostat avoids significant noise caused by the flowing cryogen. The UHV compatible cryostat [4] is mounted below the vacuum system in a superinsulation dewar which can contain a maximum of 93 liters of liquid helium (LHe) or liquid nitrogen (LN). It is equipped with a double wall between the UHV part and the cryogen bath in order to isolate the UHV part from vibrations caused by the boiling of the cryogen. The interior of the double wall may be filled with helium gas to regulate the thermal coupling of the bath to the inner wall. The design of the insert's inner tube (diameter: 120 mm) includes integrated heaters for the bakeout and a conically shaped copper band approx. 30 cm above ground. A heavy cone, made of oxygen-free, highly-conductive (OFHC) copper which carries the SFM (cf. Figs. 3.103, 3.105 and 3.106) fits exactly in it, effectively shielding the microscope from heat from above and cooling the microscope by thermal conductance.

The transfer of the microscope from the main chamber into the cryostat is per-



Figure 3.105: Principle of the vertical transfer mechanism. Left: Side view of the complete manipulator; right: front view of the double chain.

formed by the vertical transfer mechanism sketched in Fig. 3.105. Microscope and copper cone are hanging at the end of a double chain moved by two motor driven gear wheels underneath the top flange. All electrical connections to the microscope and the fiber used for the interferometric deflection detection system are held by clamps in between the chains (see the front view in Fig. 3.105). Copper beryllium meshes ensure electrical shielding and serve as a mechanical guard. The 35 electrical wires of 0.1 mm diameter are made of copper or manganine (a copper-manganese-nickel alloy with low thermal conductivity), depending on their application.

The weight of the cone and the microscope provides the force necessary to establish a good thermal contact between the cone and its counterpart in the cryostat. The thermal load to room temperature is effectively reduced by a radiation shield and by the long series of point contacts between the individual chain links if the chain hangs loose. The loose chain also decouples the cryostat mechanically from the main chamber.

A second advantage of the mechanism is that it can be built comparatively compact:

In our design, one of the rods which connects the two chains is somewhat longer than the others. This longer rod is caught by a guiding groove, which forces the chain to fold up (see Fig. 3.105). Minimal bending radii of the chain of about 2 cm are ensured. Consequently, this mechanism needs much less vertical space than a rigid transfer rod, which is important in case of limited room height. An additional advantage is that this system is significantly less fragile than transfer mechanisms which use long welded bellows.

The transfer of the microscope from the main chamber into the cryostat (≈ 1.5 m travel) or back needs about 2-5 min, depending on the selected velocity of the gear wheels. During the movement of the chain, the background pressure rises due to molecules desorbing from the chain links, but stays always below 1×10^{-9} mbar. The temperature is measured with two Si diodes [5]. One of these sensors is connected to the lower side of the cone, and the other to the bottom of the microscope. According to these sensors, the lowest temperature reached at the moment is ≈ 9 K. The microscope needs, due to the great mass of the copper cone, 10 hours for precooling with LN down to 77 K, and additional 10 hours are necessary for the final cooling with LHe down to 10 K. Operation time without refill is 3 days, enabling long-term experiments at very stable conditions. After a tip and/or sample exchange at room temperature, the microscope needs about 4 h to equilibrate.

Microscope Design

The design process of the SFM itself (Fig. 3.106) was primarily guided by its use for high-resolution atomic-scale imaging in UHV and at low temperatures. Central points were therefore UHV and LT compatibility, avoiding of misalignments during temperature changes, the reliability of all parts (piezos, motors for coarse approach, deflection detection sensor) also at low temperatures, and the mechanical stability and rigidity of the microscope.

To detect the cantilever deflection, we chose an all-fiber interferometer, which has a very high sensitivity. All-fiber interferometers have already been successfully implemented in several systems [6–11] even for the case of LT applications [7,8]. We chose a set-up which is similar to the one described in detail in Ref. [11]. In this way, we could keep all electronic components on the outside, feeding the fiber through a swagelock into the UHV system. The ≈ 3 m total travel of the light from the cantilever backside to the signal photo diode does not significantly reduce the signal quality since the light in the glass fiber is not affected by the stray fields of the electrical cables used for microscope control. Moreover, because only the fiber end has to be positioned as close as possible to the cantilever backside, the design of the whole microscope can be kept comparatively small, rigid and symmetric.

The use of a scientific instrument in UHV and in a wide temperature range (9 K to 300 K operation temperatures, 450 K during bakeout) drastically reduces the number of materials that can be used. Our microscope body (Fig. 3.106) is made from a single piece of macor [12] to keep its susceptibility to vibrations low. For most metal parts,



Figure 3.106: Schematic view of the microscope body and the copper cone. The cantilever pick-up stage (situated in the middle of the microscope body at the height of the viewports), the piezo stacks (the "legs" of the "walkers") and the cover plates of the two motors are omitted. Of the motors, only the sapphire prisms are drawn in order to show the principle mechanism of the fiber and sample coarse aproach.

titanium was chosen since it has a similar thermal contraction constant as macor, thus reducing misalignment and stress that could crack the ceramic to a minimum. One of the main problems brought up by variable temperatures in the SFM is, as already mentioned above, keeping misalignment of the cantilever relative to the fiber end to a minimum. This issue is accounted by making the microscope body as symmetric as possible: It exhibits a cylindrical overall shape with 13 cm height and 6 cm diameter and possesses exact mirror symmetry along the cantilever axis. This ensures that the fiber end does not leave the cantilever backside during cooling. Consequently, an *x-y*-coarse positioning device for the fiber end can be omitted; only fine adjusting of the fiber end is necessary which is enabled by a piezo tube which can be deflected by $\pm 3 \ \mu m$ in *z*- at room temperature.

Coarse approach of both fiber to cantilever backside (z-direction) and sample to tip is accomplished by a special type of piezo motor. It uses the lateral displacement of shear piezo stacks for the movement of a sapphire prism which travels along its long axis. It has already been described elsewhere [13] and is known to work in UHV and at temperatures well below 1 K [14]. Contrary to other piezo motors which are based on a "stick-slip" mechanism by shearing all piezos simultaneously, one piezo stack (a "leg") is sheared after the other in the present design. Thus, the motor is often called "walker". 50-300 nm travelling distance per step (depending on the operation temperature) and 50 steps/second are realized in our set-up and with the current driving electronics.

As the laser light ($\lambda = 780$ nm) is produced by a laser diode, the coherence length is quite short. This has the advantage that the interference signal is not disturbed by unwanted reflections at irregularities or at joints of the fiber, but also means that the fiber has to be brought as close to the cantilever as possible to produce a maximum interference signal. Additionally, a small cavity reduces the phase noise significantly. With the help of the coarse and the fine approach mechanisms described above ("walker" and piezo tube), we can position the fiber within a distance of 1/4 wavelength from the cantilever backside. As a result, we routinely achieve a vertical noise level below 0.1 Å peak to peak in DC-1 kHz bandwidth.

To ensure proper use of the system under UHV conditions, cantilevers and samples need to be exchanged *in situ*. The samples are mounted on small metal plates which are held on the scanner by springs; transfer is performed using a wobblestick. Cantilevers are glued to commercially available cantilever holders [15] and placed in a fixed pick-up stage in the microscope. This pick-up stage is situated on a thin piezo plate which enables the oscillation of the cantilever during dynamic mode measurements. Tunneling through the tip of the cantilever into the sample is possible; replacing the cantilever by a fixed metal tip transforms the SFM into a scanning tunneling microscope. In order to guarantee a proper alignment of the cantilever relative to the fiber end, the cantilevers are prepositioned on their holders before they are introduced into UHV, using a second pick-up stage for the holders and an optical stereo-microscope.

Performance

Figure 3.107 shows an atomic scale image of the LiF(001) surface acquired in the static contact mode at room temperature in UHV. Contrast formation in this mode strongly depends on frictional forces in the contact region, which consists of several tens or hundreds of atoms under typical experimental conditions (see Sect. 3.4.1 and Sect. 3.4.2). As a consequence, only the translational symmetry of the sample is preserved, and the periodicity in Fig. 3.107 matches the LiF lattice constant. Therefore, in this mode, only unit cell resolution can be achieved; single point defects or atomic resolution across step edges cannot be observed. The square-like appearance of the *atoms* and the triangular contrast distribution within the squares is caused by the influence of the above mentioned frictional forces (see Sect. 3.4.1 for more details).

Figure 3.108 demonstrates the capability of the dynamic non-contact mode. Here, the n-InAs(110) surface is imaged at 14 K in ultrahigh vacuum. In this surface, the arsenic atoms relax outwards and the indium atoms inwards. Therefore, usually only the arsenic atoms are visible instead of a zigzag chain of alternating indium and arsenic atoms along the $[1\bar{1}0]$ direction (see also Sect. 3.4.2). The corrugation amplitude is



Figure 3.107: LiF(001) as seen in the static contact mode at room temperature. The contrast is governed by lateral forces acting on the tip in the contact region. Parameters: The corrugation is 100 pm with a noise level of 10 pm. $F_{\rm N} \approx 7$ nN, k = 0.1 N/m, $f_{\rm res} = 12.8$ kHz, image size: 3.0 nm × 3.0 nm.



Figure 3.108: n-InAs(110) imaged in the dynamic non-contact mode at 14 K. The surface is As terminated due to a relaxation, and consequently only the arsenic atoms are visible. The corrugation amplitudes are of the order of 20 to 40 pm (depending on the crystallographic direction) with a noise level below 10 pm. Parameters: k = 36 N/m, $f_{\rm res} = 160$ kHz, $A = \pm 12.7$ nm, image size: 3.5 nm \times 1.9 nm.

about 25 pm with a noise level below 10 pm. Further examples of data obtained with this instrument at various temperatures in different modes of operation are presented in the section 3.4.2.

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Design of a cryogenic scanning force microscope for the characterization of frozen biological samples

J. H. Müller and U. D. Schwarz

Introduction

With the ongoing progress in scanning probe techniques, nanometer- and atomic-scale imaging became possible on a variety of different samples. On biological systems, however, the resolution is often limited by the intrinsic softness of the investigated materials. These drawbacks can be overcome by profiling samples which have been immobilized using cryo-techniques. For this purpose, a scanning force microscope was developed specially suited for the investigation of soft biological samples at low temperatures. The microscope is mounted on a commercially available vibration isolation system and is integrated into a UHV chamber. The vacuum system consists of a transfer and preparation chamber including a cleavage stage for operation at low temperatures and an analysis chamber with the microscope. After low-temperature cleavage, the cold sample can be exchanged with cryo-SEM as well as with cryo-TEM devices. A flow cryostat cools the sample to the temperature of liquid nitrogen, while the tip is scanned with a maximum range of 50 mm \times 50 mm. The detection system is based on the interferometer principle. In the next sections, we will discuss the functional parts of the scanning force microscope in more detail.

General Design

Figure 3.109 shows an overview of the most important parts of the SFM. The sample is mounted on a special sample holder made of Cu-Be. It is connected with a flowcryostat via gold-coated copper braids. When using liquid nitrogen as a cryogen, a temperature of approximately 120 K can be reached. The sample holder is connected to the mounting flange with an insulating glass plate and three rods. In this design, the sample is fixed in order to guarantee a good cooling; instead, the cantilever, which is attached to the end of a piezo scanner, is scanned. The coarse approach of the tip towards the sample is performed by shear piezos, which drive a slider made of sapphire in vertical direction ("z-direction") relative to the microscope body.

Stepper Motor for Coarse Approach

Figure 3.110 shows the design of the stepper motor, which uses the "walker" principle. The motor is housed in a body made of $Macor^{TM}$, a glass ceramic which is known for its high mechanical stability. With a maximum travel range of 30 mm, the motor enables the coarse approach of the tip towards the sample; it can be driven manually as well as by computer control. A slider made of sapphire is moved by applying voltage ramps to six shear piezo stacks. The stacks are sheared successively and then relaxed



Figure 3.109: Schematic of the scanning force microscope.

together, moving the slider one step forward. The driving electronics realises currently a stepping frequency of 50 Hz with a step width of 500 nm. The advantages of this motor include reliability, especially in UHV environment and at low temperatures, and excellent stability.



Figure 3.110: Principle of the steppermotor ("walker").

x-y-Translation Stage

An x-y-translation stage allows positioning of the tip in an area of 10 mm \times 10 mm. This allows simultaneous measurements of different samples, attached to the same sample holder, leading to a higher efficiency during the measurements. Three shear piezo stacks are again used to make the translation in x- and y-direction, while two guiding tracks prevent a tilting of the stage. The driving electronics is similar to the one used with the stepper motor ("walker"). In contrast to the walker, strong Cobalt-Samarium magnets hold the stage together.

Detection System for the Cantilever Deflection

Since the sample is stationary, the cantilever and the detection system have to be scanned. To reduce the mass of the detection system, we have chosen a design based on an interferometer (see Fig. 3.111). Such an all-fiber interferometer has the advantage that only the fiber end has to be scanned, allowing the scanner to be built very compact. The light of a laser diode ($\lambda = 780$ nm, $P_0 = 40$ mW) is passing a Faraday isolator (preventing backscattering of light) and is coupled to a monomode fiber. The beam is then split into a reference arm and a signal arm. The cleaved fiber end reflects up to 4% of the light, interfering with the light coming back from the cantilever backside. This interference signal is detected with a photodiode outside the UHV chamber, allowing distance sensitive measurements with a resolution in the sub-Ångstrom range.



Figure 3.111: Schematic of the glassfiber interferometer.



Figure 3.112: UHV-System with analysis- (left) and preparation chamber (middle).

Scanner

The scanner is the most complicated device in the present design. It is not only responsible for the scanning of both cantilever and detection system, but it also has to be stable enough to allow an in-situ exchange of the cantilever with an UHV manipulator. In general, the scanner consists of a concentric arrangement of two piezo tubes in which the inner piezo tube (diameter: 3.2 mm; wall thickness: 0.5 mm) holds the end of the fiber. The outer tube (diameter: 6.25 mm; same wall thickness) enables the scanning of the tip over the sample with a maximum scan range of 50 μ m × 50 μ m. At the top of the outer piezo, a special block fixes the tip holder magnetically.

Vacuum system and Preparation Chamber

The home-built scanning force microscope is attached to a commercially available eddy current vibration isolation system (see Fig. 3.109). This vibration isolation system is integrated into an UHV system consisting of the microscope chamber, a preparation chamber and the air lock (see Fig. 3.112). The microscope chamber is evacuated by both turbo and ion getter pumps. Manometers, a quadrupole mass spectrometer and a sputter gun can be connected. The last one mentioned could be useful when preparing and cleaning tips for use in non-contact measurements. In a movable carussel up to eight tips can be stored. The preparation chamber and the airlock was designed by Bal-Tec GmbH and use a separate turbo pump. A LN_2 cooled sample holder in the preparation chamber allows an in-situ freeze fracturing of the sample. The complete UHV system rests on four pneumatic damping legs to isolate the system from vibrations.

3.5.4 Magnetic Force Microscopy in Ultrahigh Vacuum

M. Dreyer, M. Kleiber, and A. Wadas

Since the invention of magnetic force microscopy (MFM) in 1987 [1] the method was mostly used under ambient conditions to study the magnetic field distribution of a surface. The technique relies on the detection of magnetic forces or force gradients exerted on a magnetically coated probing tip. The tip is scanned over the sample surface at a given tip to sample separation of typically some tens of nanometers. The forces responsible for the topographic contrast are of short range order and are negligible when scanning at a certain separation compared to the magnetic forces. Thus, an image of the magnetic field distribution of the sample is obtained. From the time of invention of MFM the performance of this technique was tested in applying it to different magnetic materials from hard magnets like FeNdB to soft ones like garnets [2–5].

Any thin interface layer present under ambient conditions between the sample and the tip changes the overall tip-sample interaction potential. This layer consists mainly of water with some other molecules like hydrocarbons from the air. This however, changes the force of adhesion necessary to apply in order to break a contact between the MFM tip and the sample. Moreover, the oxidization process might influence the magnetic sample and the thin film MFM tip while running MFM in air. Covering the sample by a protection layer doesn't help, since the layer itself may influence the magnetic structure especially of a thin film sample [6]. Ultrahigh vacuum (UHV) provides conditions to study magnetic materials by MFM without any non-magnetic layer between the sample and the probing tip. Applying MFM under UHV conditions leads to some more advantages. Since neither the magnetic probe tip nor the sample needs to be protected against oxidation the effective tip to sample distance is reduced. Furthermore, when measuring force gradients the sensitivity is enhanced. To measure force gradients the cantilever is vibrated at its resonance frequency $\omega = \sqrt{k_{\rm eff}/m}$. m denotes the mass of the cantilever and k_{eff} the effective spring constant. The latter is given by $k_{\text{eff}} = k - F'$, where F' is the overall force gradient acting on the cantilever and k is the nominal spring constant of the cantilever. The change in the force gradient while scanning over the surface shifts the resonance frequency ω . In our system the shift due to magnetic tip-sample interaction is measured with a frequency modulation (FM) detector. The lowest detectable force can be estimated to be [7]:

$$F_{\min} = \frac{1}{A} \sqrt{\frac{4kk_{\rm B}TB}{\omega_0 Q}} \tag{3.23}$$

where A denotes the rms amplitude of the oscillation, B the detection bandwidth and Q the quality factor of the oscillator. Since the Q-factor of the cantilever is increased due to the absence of air damping the sensitivity is increased when measuring in UHV.





Figure 3.113: Schematic view of the *lift mode*. In a two pass process, first the topography of a scan line is measured. The data is subsequently used to scan the line again at a given tip to sample distance.

We changed the controlling software of a commercial UHV STM/AFM provided by Omicron [8] to enable MFM measurements. We implemented two modes of operation. The first mode is displayed in figure 3.113. Each line is scanned twice. In the first pass the topography is measured. This may be performed in contact, non contact or even STM mode. For the second pass the tip is retracted from the sample. The same line is scanned again using the previously obtained topographic information to maintain a predefined tip to sample distance. The procedure is repeated for all lines of an image.

The second mode was designed to reduce the number of withdraws and approaches of the tip (cf. fig. 3.114). The sample is scanned in topographic mode along two or four sides outside the region to be scanned in MFM mode. A polynom of given order is fit to the data. The coefficients are used to extrapolate the height of the sample at each point inside the region. The first mode will be referred to as *lift mode*, the second as *plane subtraction mode* (PS-mode).

The desired height in either case is reached as follows. The feedback is switched off. The tip is retracted to a first height to overcome a strong adhesion interaction between tip and sample. At this height the oscillation of the cantilever is switched on in case we want to measure force gradients. Then the tip is lowered to the final height at which the sample will be scanned.

Each mode has specific advantages and detriments with respect to the other. Applying the lift mode gives well related topographic and magnetic information in one



Figure 3.114: Schematic view of the *plane subtraction mode*. First the sample is scanned several times along two or four lines outside the region of interest. The scan lines of each side are averaged and a polygon of given order is fit to each of the averaged line. The parameters are used to estimate the position of the sample in 3D space and thus to keep the tip at the given distance to the sample.

measurement. Furthermore, this mode takes local topographic features into account allowing to scan at a lower tip to sample separation. For thin films prepared in UHV the local topographic features are often not of interest. Here, the PS-mode is more appropriate because it is faster as each line is only scanned once. This mode is also less dependent on the quality of topographic line scans since the data are not used directly for the height determination. This enables the use of sensors which give a good magnetic but a poor topographic signal.

Test Samples

We used samples which were already measured under ambient conditions, to test the MFM modes and the tip preparation in UHV. We have prepared our own MFM sensors by depositing iron films in the UHV chamber on commercially available Si cantilevers with integrated tips. The data presented here have been obtained with cantilevers specified to have a spring constant of 1–4 N/m and a resonance frequency of 60–80 kHz. The cantilevers were cleaned by Ar^+ –ion sputtering for 90 min, at a kinetic energy of 1.5 keV for the Ar^+ –ions. Subsequently we have used electron beam evaporation in order to deposit Fe films of the thickness of 20 nm upon the tips.

Due to ideal UHV conditions (pressure $5 \cdot 10^{-11}$ Torr) and in-situ prepared MFM tips we have been able to reach domain wall contrast on all of the test samples. The samples were a Co/Pt[15×Co(3.5 Å)/Pt(6 Å)] multilayer structure, the high coercivity thin plate of BaFe₁₂O₁₉ single crystal and a soft (YSmLuCa)₃(FeGe)₅O₁₂ magnetic thin film garnet with a bubble domain structure. On magnetic garnet (of the thickness of 4.7 μ m) we would like to observe Bloch lines and their positions under influence of an applied field. These problems are important for vertical Bloch line memories. Co/Pt multilayers are known as perpendicular recording media. Finally, $BaFe_{12}O_{19}$ single crystal is representative for a class of materials with a high anisotropy and very thin domain walls (ca. 20 nm).

In Fig. 3.115 we show the results. The first MFM image represents the magnetic force gradient detected above the Co/Pt multilayer. The domains are meander like and their thickness varies approximately from 70 nm to 160 nm. The coercivity of the MFM tip is smaller than the stray field from the domain wall. Therefore the magnetization within most of the tip volume fixes the direction in position to lie along the domain wall stray field. This leads to an attractive force measured by MFM which is represented by the black colour in the image. The measured frequency shift is about 1.5 Hz. This corresponds to a magnetic force gradient of $2 \cdot 10^{-4}$ N/m for a spring constant of the cantilever of 4 N/m and a resonance frequency of 65 kHz.

The second image measured on the $BaFe_{12}O_{19}$ single crystal shows a strong domain wall contrast corresponding to a 13 Hz resonance frequency shift. This corresponds to a magnetic force gradient of $1.6 \cdot 10^{-3}$ N/m. The width of a single domain seen as a bright stripe is approximately 700 nm.

The final MFM image shows the magnetic stray field of a thin film $(YSmLuCa)_3(FeGe)_5O_{12}$ garnet. The maximum of the shift of the resonance frequency



Figure 3.115: Domain wall contrast on three test samples imaged with *in situ* prepared magnetic tips. Left to right: $Co/Pt[15 \times Co(3.5 \text{ Å})/Pt(6 \text{ Å})]$ multilayer structure, $BaFe_{12}O_{19}$ single crystal and $(YSmLuCa)_3(FeGe)_5O_{12}$ magnetic thin film garnet.

has been 0.24 Hz over the imaged surface area. There is a cylindrical domain in the center of the image seen as a bright spot encircled by a dark ring. Other dark rings are only partly visible. The dark ring visualizes the domain wall. Disturbances within rings are observed. They happen reproducibly on the very small area of the domain wall, seen as a tip change. Since the MFM tip is nominally 140 nm appart from the surface it might occur due to a rapid change of the stray field above the Bloch line.

We found as a very promising result that by using in-situ coated MFM tips we could reveal the domain wall contrast on different magnetic samples from very soft garnet films to very hard $BaFe_{12}O_{19}$ single crystals. Thin film tips prepared in UHV are magnetically very stable and we have not seen any decrease of the MFM signal in time.

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3.6 STM Theory

3.6.1 Prediction of Corrugation Reversal in STM-Images of bcc-(110) Transition Metal Surfaces

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In the last years, the scanning tunneling microscope (STM) has developed into a powerful tool to investigate surfaces in real space and has become a driving force in the field of nanofabrication. The interpretation of STM-images is still non-trivial though, especially if atomic resolution is achieved. The reason is the exploited physical effect itself, i.e. tunneling between a (metallic) tip and a (conducting) sample, which makes it often quite difficult to disentangle the geometric and electronic structure. General wisdom, found in modern textbooks on surface physics, claims that the interpretation should be unproblematic on metal surfaces. Electrons in metals screen the nuclear charge and follow the atomic structure in a good approximation. Thus the current measured by STM should be highest on top of atom positions.

This view might hold for simple metals such as Au, Al or Cu. We show that in



Figure 3.116: Dependence of the corrugation amplitude Δz on the bias voltage of the calculated STM-image of W(110). Insets show the surface unit-cell (bottom right) with atomic sites marked by dots and typical STM-images calculated for different bias voltages V. The full line displays $\Delta z = z_{max} - z_{min}$ between the lateral points \mathbf{r}_{\parallel} of maximum and minimum corrugation within the unit cell. The dashed line displays $\Delta z = z_{atom} - z_{hollow}$, the corrugation measured between the position of the atom and the hollow site. All results are calculated at a distance of $z_0 = 4.6$ Å. Area of dark grey scale means small tunneling current.

the case of transition metal surfaces the interpretation is far from being trivial and can be strongly voltage-dependent. Our calculations of the electronic structure of the W(110)-surface, based on density-functional theory in the local density approximation and performed with the full-potential augmented plane wave method (FLAPW) in film geometry, indicate that the STM-image changes drastically with bias voltage. The described effect makes it difficult to decide upon the position of the surface atoms by STM and becomes particularly important if one tries to deduce adsorption sites on a transition metal surface from STM-images.

The result of the calculation is displayed in Fig. 3.116. The corrugation amplitude Δz as a function of the applied bias voltage V is shown. At negative bias voltages, i.e. tunneling from occupied sample states, below a certain value (-0.4 V) the hollow sites of the unit cell are imaged as protrusions while atom positions appear as dips. This unusual appearance is called anticorrugation. Choosing bias voltages above the critical value results in a change to a normal image, i.e. atom positions now coincide with high tunneling current. Thus a corrugation reversal from anticorrugation to normal corrugation with bias voltage has occured. Notice that at a voltage close to -0.4 V the calculated STM-image resembles neither of the extreme cases (normal or anticorrugation) but displays a pattern of bent stripes parallel to the short side of the rectangular unit cell.

Next we would like to understand the corrugation reversal on the basis of the electronic structure. The tunneling current is calculated in the approximation of Tersoff and Hamann. Within this model the tunneling current I as a function of the bias voltage V is given by:

$$I(\mathbf{r}_{\parallel}, z; V) \propto \int_{E_F}^{E_F + eV} n(\mathbf{r}_{\parallel}, z|\epsilon) \, d\epsilon \,.$$
(3.24)

Thus it depends exclusively on the local density of states (LDOS) of the sample

$$n(\mathbf{r}_{\parallel}, z | \epsilon) = \sum_{\mathbf{k}_{\parallel}\nu} \delta(\epsilon - \epsilon_{\mathbf{k}_{\parallel}\nu}) \sum_{n} n_{\mathbf{k}_{\parallel}\nu}^{n}(z) \exp\left(i\mathbf{G}_{\parallel}^{n}\mathbf{r}_{\parallel}\right)$$
(3.25)

at the position of the STM-tip $(\mathbf{r}_{\parallel}, z)$. The total current is evidently given by the n = 0 part $(\mathbf{G}_{\parallel}^{0} = 0)$ while the corrugation depends on terms with n > 0 $(\mathbf{G}_{\parallel}^{n>0} \neq 0)$. The local density of a certain state, normal to the surface, $n_{\mathbf{k}_{\parallel}\nu}^{n}(z)$ can be expanded to a good approximation as

$$n_{\mathbf{k}_{\parallel}\nu}^{n}(z) \approx \sum_{n',n''} c_{\mathbf{k}_{\parallel}\nu}^{n'} c_{\mathbf{k}_{\parallel}\nu}^{n'*} \exp\left[-(\kappa_{\mathbf{k}_{\parallel}\nu}^{n''} + \kappa_{\mathbf{k}_{\parallel}\nu}^{n'})z\right] \delta(\mathbf{G}_{\parallel}^{n''} - \mathbf{G}_{\parallel}^{n'}, \mathbf{G}_{\parallel}^{n})$$
(3.26)

whereas $\kappa_{\mathbf{k}_{\parallel}\nu}^{n} = \sqrt{2m|\epsilon_{\mathbf{k}_{\parallel}\nu}|/\hbar^{2} + (\mathbf{k}_{\parallel} + \mathbf{G}_{\parallel}^{n})^{2}}$ describes the exponential decay into the vacuum and the expansion coefficients of the wavefunctions, $c_{\mathbf{k}_{\parallel}\nu}^{n}$, contain the information on the electronic structure. The tunneling current is maximal with minimal $\kappa_{\mathbf{k}_{\parallel}\nu}^{n}$, i.e. important are states with minimal tunneling barrier $\epsilon_{\mathbf{k}_{\parallel}\nu}$ and (at fixed energy) states with a small value of $|\mathbf{k}_{\parallel} + \mathbf{G}_{\parallel}^{n}|$ (n > 2 can already be neglected). This is



Figure 3.117: a) Two-dimensional Fermi surface of W(110). Open circles mark bulk states while filled circles represent surface states or resonances. b) Surface band structure in $\bar{\Gamma}\bar{S}$ direction. The surface state band (open circles), appearing in the Fermi surface around the \bar{S} point, is clearly visible as well as the surface resonances around $\bar{\Gamma}$. c) and d) show $\sum_{n',n''} \exp\left[-(\kappa_{\mathbf{k}_{\parallel}\nu}^{n''} + \kappa_{\mathbf{k}_{\parallel}\nu}^{n'})z\right]\delta(\mathbf{G}_{\parallel}^{n''} - \mathbf{G}_{\parallel}^{n'}, \mathbf{G}_{\parallel}^{n})$ at $\epsilon_{\mathbf{k}_{\parallel}\nu} = \epsilon_{F}$ for n = 0 and n = 1. e) and f) additionally contain the electronic structure of W(110) given by the term $c_{\mathbf{k}_{\parallel}\nu}^{n''} c_{\mathbf{k}_{\parallel}\nu}^{n'*}$. The bias voltage is taken as -0.2 V for e) and f). Yellow and red mark positive and negative values, respectively. For the n = 1 term, positive values indicate a normal image and negative values anticorrugation.

analyzed in Fig. 3.117, showing the dependence of the contribution to the tunneling current on the \mathbf{k}_{\parallel} vector in the case of the W(110)-surface. Fig. 3.117c and d display the contribution taking the distance-dependence of the wavefunctions into account, according to $\sum_{n',n''} \exp\left[-(\kappa_{\mathbf{k}_{\parallel}\nu}^{n''} + \kappa_{\mathbf{k}_{\parallel}\nu}^{n'})z\right]\delta(\mathbf{G}_{\parallel}^{n''} - \mathbf{G}_{\parallel}^{n'}, \mathbf{G}_{\parallel}^{n})$ at $\epsilon_{\mathbf{k}_{\parallel}\nu} = \epsilon_F$ for n = 0 and n = 1. One can conclude that only states around $\bar{\Gamma}$ are important for the n = 0 term, but the most important term for the corrugation, n = 1, is dominated by states near \bar{S} . If one includes the particular electronic structure given by the term $c_{\mathbf{k}_{\parallel}\nu}^{n''}c_{\mathbf{k}_{\parallel}\nu}^{n''}$ one ends up with Fig. 3.117e and f. In the case of the n = 0 term one observes that only states from the Fermi surface close to $\bar{\Gamma}$ contribute (Fig. 3.117a) as predicted by the distance-dependence of the wavefunctions (Fig. 3.117c). All values are strictly positive since the n = 0 term is related to the local charge density. The n = 1 term, responsible for the corrugation, on the other hand displays contributions also from the surface state surrounding \bar{S} . The values of the surface state are negative and cause the anticorrugation image. Its contribution competes with positive values (indicating a normal STM-image) of bulk and surface resonance states located in the center of the Brillouin zone. Varying the bias voltage changes the weight of the competing contributions and this causes the corrugation reversal. The described corrugation reversal was also found in the case of Ta(110) and ferromagnetic Fe(110) and we believe it is a quite general phenomenon for transition metal surfaces. For example, the well studied Fe(001)-surface has a surface resonance which also causes an anticorrugation image in STM and this has already been experimentally observed. For a more detailed discussion of the effect we refer to Ref. [1].

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[1] S. Heinze, S. Blügel, R. Pascal, M. Bode and R. Wiesendanger, to be published in Phys. Rev. B.

Chapter 4 Collaborations

4.1 Research Partners

- MARCH, Hamburg: Prof. R. Anton, Prof. W. Hansen, Prof. D. Heitmann, Prof. J. Kötzler, Prof. U. Merkt, Dr. D. Grundler, Dr. T. Matsuyama
- HASYLAB, Hamburg: Prof. R. L. Johnson
- Institute for Physical Chemistry, University of Hamburg: Prof. H. Weller
- Beiersdorf AG, Hamburg: Dr. D. Emeis, Dr. U. Hintze
- Forschungszentrum Jülich: Dr. S. Blügel
- University of Bielefeld: Prof. G. Reiss
- University of Münster: Prof. H. Fuchs
- University of Saarbrücken: Prof. U. Hartmann
- Siemens, Erlangen: Dr. H. A. M. van den Berg
- PTB Braunschweig: Dr. M. Albrecht

- EPF Lausanne, Switzerland: Prof. F. Lévy
- University of Nijmegen, The Netherlands: Prof. H. van Kempen
- University of Dublin, Trinity College, Ireland: Prof. J. M. D. Coey, Dr. I. V. Shvets
- Institute of Physics, Polish Academy of Sciences, Warsaw, Poland: Prof. J. Rauluszkiewicz
- Ukrainian Academy of Sciences, Kiev, Ukraine: Prof. S. Nepijko
- University of Reijkjavik, Iceland: Prof. V. Gudmundsson
- Weizmann Institute of Science, Rehovot, Israel: Dr. Y. Manassen

4.2 Industrial Partners

- Beiersdorf, Hamburg
- Digital Instruments, Mannheim
- Exabyte Magnetics, Nürnberg
- Focus, Hünstetten–Görsroth
- IBM Deutschland Speichersysteme, Mainz
- Nanosensors, Aidlingen
- OMICRON Vakuumphysik, Taunusstein
- Oxford Instruments, Cambridge
- Siemens, Erlangen
- Siemens, München
- Surface Imaging Systems, Herzogenrath

Chapter 5

Theses

5.1 Diploma Theses

1. Christian Witt (1996):

Aufbau eines UHV-Rastertunnelmikroskops zur in-situ Untersuchung der Wachstumsstadien ultradünner Schichten

2. Hendrik Hölscher (1996):

Theorie der Reibungskraftmikroskopie

3. Jörn Asmussen (1996):

Entwicklung und Aufbau eines Rasterthermomikroskops

4. Axel Born (1996):

Analyse und Erzeugung von Nanostrukturen mittels Rasterkapazitätsmikroskopie

5. Sven Hanszen (1996):

Untersuchungen zur Rasterkraftmikroskopie an Liposomen

6. Martin Winzer (1996):

Herstellung und Charakterisierung von Nano-Dot-Arrays und Quantenpartikeln

- 7. Jens Müller (1996):Aufbau eines Rasterkraftmikroskops
- 8. Oliver Zwörner (1996):Tribologische Eigenschaften der Kohlenstoffmodifikationen

9. Christoph Zarnitz (1996):

Rastertunnelmikroskopie und -spektroskopie von dünnen Gadoliniumfilmen auf $\mathrm{W}(110)$

10. Stefan Heinze (1997):

Ab-initio Elektronenstruktur
rechnungen im Hinblick auf RTM-Experimente an ${\rm Fe}/{\rm W}(110)$

11. Michael Kleiber (1997):

Untersuchungen der magnetischen Eigenschaften von Kobaltpartikeln mittels Magnetkraftmikroskopie

12. Thomas Blum (1997):

Dünnfilm-Thermoelemente im Rasterthermomikroskop und Messung des magnetischen Streufeldes stromdurchflossener Leiterbahnen

- 13. Wolfgang Kresse (1997):UHV-Rastertunnelmikroskopie an der InAs(110)-Spaltfläche
- 14. Daniel Haude (1997):

Ultrahochvakuum-Rastertunnelmikroskopie: Instrumentierung und Anwendung

15. Alexander Kelch (1997):

Untersuchungen zur Mikrostruktur des Haares und Nachweis von Produktwirkung mittels Rasterkraftmikroskopie

16. Kai Schnee (1997):

Nahfeldoptik und der Magnetooptische Kerr Effekt

17. Matthias Münz (1997):

Aufbau eines nahfeldoptischen Mikroskops in Reflexionsgeometrie

18. Felix Kümmerlen (1997):

Mikromagnetische Simulation von Magnetkraftmikroskopieexperimenten und Ummagnetisierungsprozessen magnetischer Nanopartikel

19. Oswald Pietzsch (1997):

Aufbau eines Rastertunnelmikroskops für den Einsatz im Ultrahochvakuum, bei tiefen Temperaturen und in hohen Magnetfeldern

20. Uwe Christian Mick (1997):Untersuchungen zur Homoepitaxie auf der Cr(001)-Oberfläche

21. Heinz Tödter (1997):

RTM und RTS an ultradünnen Filmen von Terbium, Gadolinium-Terbium und Gadolinium-Eisen

22. Martin Hennefarth (1998):

Wachstum und elektronische Eigenschaften ultradünner Manganfilme auf Wolfram(110)

23. Shenja Langkat (1998):

Untersuchungen zur Elastizitätsmessung mittels Rasterkraftmikroskopie und Aufbau eines Tieftemperatur–Ultrahochvakuum–Rasterkraftmikroskops mit einem Probeverschiebetisch

24. Roland Schneider (1998):

Aufbau eines UHV–Rastertunnelmikroskops und Anwendung auf das Wachstum dünner Schichten auf InAs(110)

5.2 Ph. D. Theses

1. Hendrik Bluhm (1996):

Untersuchung ferroelektrischer Kristalle mit dem Rasterkraftmikroskop

2. Matthias Bode (1996):

Strukturelle und lokale elektronische Eigenschaften ultradünner Eisenfilme auf W(110)

3. Kirsten Ingolf Schiffmann (1997):

Mikrostruktur und Mikrotribologie von amorphen metallhaltigen Kohlenwasserstoffschichten

4. Markus Löhndorf (1997):

Untersuchung mikromagnetischer Eigenschaften dünner Schichten mit dem Magnetkraftmikroskop

5. Ralph Dombrowski

Aufbau eines Ultrahochvakuum-Tieftemperatur-Rastertunnelmikroskops mit rotierbarem Magnetfeld und magnetfeldabhängige Rastertunnelspektroskopie an der InAs(110)-Oberfläche

6. Christian Wittneven (1998):

Aufbau eines Ultrahochvakuum-Tieftemperatur-Rastertunnelmikroskops mit drehbarem Magnetfeld und Untersuchung der Streuzustände ionisierter Dotieratome an InAs(110)

7. Alexander Schwarz (1998):

Aufbau eines Ultrahochvakuum–Tieftemperatur–Rasterkraftmikroskops und Untersuchung des Abbildungsmechanismus auf LiF(001) und InAs(110)

8. Wolf Allers (1998):

 $\label{eq:constraint} Tieftemperatur-Rasterkraftmikroskopie \ \mbox{im} \ Ultrahochvakuum \ \mbox{auf} \ \mbox{van}-der-Waals-Oberflächen$

9. René Pascal (1998):

Rastertunnelmikroskopie und Rastertunnelspektroskopie an dünnen Filmen der Seltenerdmetalle Gd und Tb sowie $GdFe_2$ Legierungen

10. Michael Dreyer (1998):

Untersuchungen ultradünner Kobaltfilme mittels Magnetkraftmikroskopie im Ultrahochvakuum

11. Hendrik Hölscher (1998):

Kontrastmechanismen in der Rasterkraftmikroskopie

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6.1 Books

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- 62. M. Getzlaff, M. Bode, and R. Wiesendanger, Phys. Rev. B **58** 9681 (1998): Determination of radial matrix elements and phase shifts in the photoemission process using a rotatable electric field vector.
- 63. M. Getzlaff, M. Bode, A. Wadas, and R. Wiesendanger, Proc. ICEM14, Cancun (Mexico), Electron Microscopy Vol. III, 143 (1998): Recent developments in scanning probe microscopy and spectroscopy for imaging of magnetic domains.
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- 66. A.A. Bukharaev, D.V. Ovchinnikov, N.I. Nurgazizov, E.F. Kukovitskii, M. Kleiber, and R. Wiesendanger, Physics of the Solid State **40** 1163 (1998): Investigation of micromagnetism and magnetizatic reversal of Ni nanoparticles using a magnetic force microscope.
- 67. A.A. Bukharaev, D.V. Ovchinnikov, N.I. Nurgazizov, E.F. Kukovitskii, M. Kleiber, and R. Wiesendanger, Proc. 6th Int. Symp. Nanostructures: Physics and Technology, St. Petersburg, Russia, p. 428 (1998): Micromagnetic properties and magnetization reversal of Ni nanoparticles studied by magnetic force microscopy.
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- 70. M. Bode, M. Getzlaff, and R. Wiesendanger, Phys. Rev. Lett. 81 4256 (1998): Spin-Polarized Vacuum Tunneling into the Exchange-split Surface State of Gd(0001).

- 71. R. Wiesendanger, M. Bode, M. Getzlaff, M. Kleiber, M. Löhndorf, R. Pascal, A. Wadas, and Ch. Zarnitz, Scanning Microscopy (in press): Formation and Physical Properties of Magnetic Nanostructures studied by Scanning Probe Spectroscopy.
- 72. S. Heinze, S. Blügel, R. Pascal, M. Bode, and R. Wiesendanger, Phys. Rev. B (in press): Prediction of bias-voltage dependent corrugation reversal for STM images of bcc-(110)-surfaces: W(110), Ta(110) and Fe(110).
- 73. A. Born and R. Wiesendanger, Appl. Phys. A (in press): Present and future developments of SPM systems as mass storage devices.
- 74. M. Dreyer, M. Kleiber, A. Wadas, and R. Wiesendanger, Phys. Rev. B (in press): Composition driven change of magnetic anisotropy of ultrathin Co/Au (111) films studied by magnetic force microscopy under ultrahigh vacuum.
- 75. Ch. Wittneven, R. Dombrowski, M. Morgenstern, and R. Wiesendanger, Phys. Rev. Lett. (in press): Scattering States of Ionized Dopants probed by Low Temperature Scanning Tunneling Spectroscopy.
- 76. H. Hölscher, W. Raberg, U.D. Schwarz, A. Hasbach, K. Wandelt, and R. Wiesendanger, Phys. Rev. B (in press): *Imaging of sub-unit-cell structures in the* contact mode of the scanning force microscope.
- 77. W. Allers, A. Schwarz, U.D. Schwarz, and R. Wiesendanger, Appl. Surf. Sci. (in press): Dynamic scanning force microscopy at low temperatures on a van der Waals surface: graphite (0001).
- 78. A. Schwarz, W. Allers, U.D. Schwarz, and R. Wiesendanger, Appl. Surf. Sci. (in press): Simultaneous imaging of the In and As sublattice on InAs(110)-(1x1) with dynamic scanning force microscopy.
- 79. H. Hölscher, U.D. Schwarz, and R. Wiesendanger, Appl. Surf. Sci. (in press): Calculation of the frequency shift in dynamic force microscopy.
- 80. M. Bode, M. Getzlaff, S. Heinze, and R. Wiesendanger, Appl. Surf. Sci. (in press): New insight into the surface magnetic properties of Gd(0001).
- 81. M. Getzlaff, R. Pascal, H. Tödter, M. Bode, and R. Wiesendanger, Appl. Surf. Sci. (in press): *GdFe*₂ alloy formation observed by *STM*.
- 82. M. Getzlaff, M. Bode, R. Pascal, and R. Wiesendanger, Appl. Surf. Sci. (in press): The adsorption process of hydrogen on Gd(0001).
- 83. M. Getzlaff, M. Bode, and R. Wiesendanger, Appl. Surf. Sci. (in press): Coadsorption of H and CO on Gd(0001).

- 84. R. Dombrowski, Chr. Steinebach, Chr. Wittneven, M. Morgenstern, and R. Wiesendanger, Phys. Rev. B (in press): Determining the tip induced band bending by scanning tunneling spectroscopy of the states of the tip induced quantum dot on InAs(110).
- 85. M. Getzlaff, M. Bode, R. Pascal, and R. Wiesendanger, Phys. Rev. B (in press): Adsorbates on Gd(0001) - a combined STM/UPS study.
- 86. W. Allers, A. Schwarz, U.D. Schwarz, and R. Wiesendanger, submitted to Europhys. Lett.: Dynamic scanning force microscopy at low temperatures on a noble gas crystal: Atomic resolution on the xenon (111) surface.
- 87. M. Bode, M. Getzlaff, and R. Wiesendanger, submitted to J. Vac. Sci. Technol.: Quantitative Aspects of Spin-Polarized Scanning Tunneling Spectroscopy of Gd(0001).
- 88. M. Bode, M. Hennefarth, D. Haude, M. Getzlaff, and R. Wiesendanger, submitted to Surf. Sci.: Growth of thin Mn-films on W(110) studied by means of in-situ scanning tunneling microscopy.
- 89. R. Wiesendanger, M.Bode, and M. Getzlaff, submitted to J. Magn. Soc. Jpn.: Recent advances in spin-polarized scanning tunneling spectroscopy for imaging of magnetic domains.
- 90. A. Pundt, M. Getzlaff, M. Bode, R. Kirchheim, and R. Wiesendanger, Phys. Rev. Lett. (submitted): Nano-scale modifications during hydrogen absorption in Gd thin films.

Chapter 7 Talks

7.1 Invited Talks

- **22.02.96:** R. Wiesendanger, Enschede, The Netherlands (CMO Symposium): Observation and local spectroscopy of metallic nanostructures.
- **12.03.96:** R. Wiesendanger, Trieste, Italy (2nd Workshop on Science and Technology of Thin Films): *Nanoelectronic properties of thin films*.
- 14.03.96: R. Wiesendanger, Trieste, Italy (2nd Workshop on Science and Technology of Thin Films): Nanomagnetic properties of thin films.
- 14.03.96: R. Wiesendanger, Trieste, Italy (2nd Workshop on Science and Technology of Thin Films): Nanomechanical properties of thin films.
- **25.03.96:** R. Wiesendanger, Tsukuba, Japan (1st Int. Symposium on Advanced Physical Fields): Local and Scanning Tunneling Spectroscopy of 0D - 3D Metallic Nanostructures.
- **08.05.96:** R. Wiesendanger, MPI für Festkörperforschung, Stuttgart, Germany, (Workshop on "Nanoscale Science with Scanning Probe Techniques"): Scanning Probe Spectroscopy of Magnetic Nanostructures: from single atoms to 3D-islands.
- **29.05.96** R. Wiesendanger, Atlanta, USA (40th Int. Conf. Electron, Ion and Photon Beam Technology and Nanofabrication): Application of Scanning Probe Methods for Electronic and Magnetic Device Fabrication, Characterization and Testing.
- **21.06.96:** R. Wiesendanger, Fraunhofer-Institut für Siliziumtechnologie, Itzehoe, Germany (1. Norddeutscher Technologiedialog): Von der Mikro- zur Nanotechnik mit Spitzentechnologien.
- **10.09.96:** R. Wiesendanger, Peking, China (NANO-IV): Magnetic Nanostructures studied by Scanning Probe Microscopy and Spectroscopy.

- 10.09.96 U. D. Schwarz, H. Bluhm, H. Hölscher, O. Zwörner, P. Köster, W. Allers, A. Schwarz, and R. Wiesendanger. Beijing, China (Fourth International Conference on Nanometer-scale Science and Technology (NANO IV)): Experimental and theoretical aspects of point contact friction.
- **07.11.96:** R. Wiesendanger, Bad Honnef, Germany (WE-Heraeus-Seminar über Nanostrukturwissenschaften und -technik): *Beiträge der Rastertunnel-Mikroskopie* und -Spektroskopie zur Nanostrukturwissenschaft.
- **27.01.97:** R. Wiesendanger, Bad Honnef, Germany, (WE-Heraeus-Seminar "New Nanostructures below 100 nm: Perspectives and Applications"): Nanofabrication by Scanning Probe Methods: Current Status and Challenges for the Future.
- **22.03.97:** R. Wiesendanger, Großarl, Austria, (Seminar "Grundlagen und Technologie elektronischer Bauelemente"): Rastersondenmikroskopie und -spektroskopie an Halbleiteroberflächen und Bauelementen.
- **25.03.97:** R. Wiesendanger, MNU'97, Hamburg, Germany: Aktuelle Entwicklungen in der Mikro- und Nanostrukturforschung.
- **13.05.97:** R. Wiesendanger, Chicago, USA (Scanning Microscopy '97) Formation and Physical Properties of Magnetic Nanostructures studied by Scanning Probe Spectroscopy.
- **20.06.97:** R. Wiesendanger, Munich, Germany (3rd German-American Annual Symposium "Frontiers of Science"): Nano-Science and Nano-Technology: An Excursion into a Small World.
- **15.07.97:** U. D. Schwarz, O. Zwörner, and H. Hölscher, First International Symposium on Scanning Probe Spectroscopy and Related Methods, Poznan, Poland: *Quantitative friction force spectroscopy and the nature of the tip/sample contact in force microscopy*.
- 17.07.97: M. Bode, R. Pascal, Ch. Zarnitz, M. Getzlaff, S. Heinze and R. Wiesendanger, SPS'97, Posnan (Poland): Scanning Tunneling Spectroscopy of thin metal films on W(110): correlation of structural and local electronic properties.
- 28.07.97: A. Born, and R. Wiesendanger, First German-Chinese Workshop on Fundamentals of Nanometer-Scale Structuring: Applications of Scanning Probe Microscopy and Selforganisation Processes, Karlsruhe, Germany: Present and future developments of SPM-systems as mass storage devices.
- **19.09.97:** R. Wiesendanger, Haydau, Germany (Workshop on Nanostructures): Correlation between nano-scale structural, electronic and magnetic properties of thin films by scanning probe microscopy and spectroscopy.

- **06.11.97:** U. D. Schwarz, H. Hölscher, H. Bluhm, O. Zwörner, and R. Wiesendanger; 2nd Seminar on Quantitative Microscopy, (QM2), Vienna, Austria : *Tip-sample interaction in contact force microscopy*.
- 12.11.97: M. Bode and R. Wiesendanger, KEMIA'97, Helsinki, Finland: Recent Developments in SPM Techniques.
- **11.12.97:** R. Wiesendanger, Kanazawa, Japan (5th Int. Coll. on STM): Local electronic properties in the presence of internal and external magnetic fields studied by variable-temperature scanning tunneling spectroscopy.
- 17.02.98: R. Wiesendanger, Bad Honnef, Germany (WE-Heraeus-Seminar on Spindependent Transport): Nanoscopic Spin-Valve Structures.
- 17.03.98 M. Bode, M. Getzlaff, R. Pascal, and R. Wiesendanger, Wandlitz, Germany (190. WE-Heraeus-Seminar): Temperature-dependent scanning microscopy and spectroscopy of Gd layers on W(110).
- **19.03.98:** R. Wiesendanger, Regensburg, Germany (TATF98): Correlation between atomic, local electronic and magnetic structure of thin films by scanning probe spectroscopy.
- **24.03.98:** M. Morgenstern, Regensburg (DPG-Frühjahrstagung): Tieftemperatur-Rastertunnelspektroskopie an InAs(110): Landauniveaus und Streuung von Elektronenwellen an Dotieratomen.
- **25.03.98:** A. Born, Düsseldorf, (InCom 98): Rastersondenmikroskopie: spezielle Anwendungen in der Mikroelektronik.
- **29.03.98:** M. Morgenstern, Dallas, Texas, USA (215th American Chemical Society National Meeting): Scanning tunneling microscopy investigation of H_2 O-adsorption on Pt(111).
- **22.04.98** M. Bode, M. Getzlaff, A. Wadas, and R. Wiesendanger, Rehovot, Weizmann-Institute, Israel (IVS 97): Recent developments on scanning probe microscopy and spectroscopy of thin magnetic films.
- **06.05.98** U. D. Schwarz, H. Hölscher, O. Zwörner, and R. Wiesendanger. Bad Honnef (196th Seminar of the WE-Heraeus Foundation on Mathematical Modelling and Analysis of Nonsmooth Dynamical Systems): *Nanotribology with the scanning* force microscope.
- 11.05.98: R. Wiesendanger, La Grande Motte, France (ECSC-6): Scanning Tunneling Microscopy and Spectroscopy: Correlation of Structural, Electronic and Magnetic Information on the Nanometer Scale.

- 16.05.98: R. Wiesendanger, University of Heidelberg, Germany (Int. Symp. on Modern Topics in Applied Physics): Low-Temperature Scanning Tunneling Spectroscopy of Electronic States in Semiconductors and Ferromagnets.
- **18.05.98:** M. Morgenstern, Hindsgval, Denmark (Nanoscience for Nanotechnology): Recent developments in scanning probe instrumentation.
- **26.05.98:** M. Morgenstern, Jerusalem, Israel (Minerva / GIF-Center Symposium): Local electronic properties in the presence of internal and external magnetic fields studied by variable temperature scanning tunneling spectroscopy.
- **29.05.98:** R. Wiesendanger, Kraków, Poland (Int. Symp. Nano-Scale Modification of Surfaces: *Characterisation and Modification of Magnetic Nanostructures by Local Probe Methods*.
- **29.06.98:** M. Morgenstern, Schloß Ringberg (Ringberg Workshop on Microfabrication, Nanostructured Materials and Biotechnology): Nanostructural investigations by scanning probe microscopy.
- 10.07.98 U. D. Schwarz, Technical University of Berlin (3rd meeting of the working group "Thin films in microsystem technology", German Society of Material Science, thin film division): Film thickness measurement by means of scanning force microscopy (German).
- **17.07.98:** A. Born, Cambridge, UK (Opening ceremony of the D.I. Technical Centre and Sales Office): *Applications of scanning capacitance microscopy*.
- **30.07.98:** M. Morgenstern, Montpellier, France (8th International Conference on Shallow Level Centers in Semiconductors SCLS98): Scanning tunneling spectroscopy on InAs(110): scattering of electron waves at ionized dopants.
- **02.09.98:** M. Getzlaff, M. Bode, A. Wadas, and R. Wiesendanger. Cancun, Mexico (ICEM-14): Recent developments in scanning probe microscopy and spectroscopy for imaging of magnetic domains.
- **22.09.98:** M. Morgenstern, London, Great Britain (EMAG Workshop): Scanning tunneling spectroscopy on InAs(110): scattering of electrons at ionized dopants and Landau quantization.
- **23.10.98:** R. Wiesendanger, Nijmegen, The Netherlands (6th Dutch Symposium on Scanning Probe Microscopy): Variable-temperature scanning tunneling spectroscopy on semiconductors and ferromagnets.
- **29.10.98:** M. Morgenstern, Hamburg (II. SFB-Workshop on Quantum Materials, V. Hamburger Symposium on the Physics of Micro- and Nanostructures): Scanning tunneling spectroscopy on InAs(110): scattering of electron waves at ionized dopants and spatially resolved Landau level spectroscopy.

7.2 Conference Contributions and Talks at Other Institutes

- **04.01.96:** M. Bode, R. Pascal, and R. Wiesendanger, Bad Honnef, Germany (Dreikönigstreffen): Atomic and local electronic structure of thin Fe films on W(110) studied by scanning tunneling microscopy/spectroscopy.
- **05.01.96** M. Löhndorf, A. Wadas, and R. Wiesendanger, Bad Honnef, Germany, (Oberflächenmagnetismus): *Magnetic force microscopy study of thin magnetic Films*.
- **10.01.96:** R. Wiesendanger, IFW Dresden, Germany (Kolloquium): Korrelation struktureller, elektronischer und magnetischer Eigenschaften auf der Nanometerskala mittels Rastersondenmikroskopie und -spektroskopie.
- 18.01.96: R. Wiesendanger, University of Kassel, Germany (Kolloquium): Rastertunnelmikroskopie und Spektroskopie an einzelnen Adatomen, Clustern, 2D- und 3D-Inseln auf metallischen Substraten.
- **29.01.96:** R. Wiesendanger, University of Ulm, Germany (Kolloquium): Rastersondenmikroskopie und Spektroskopie: Zugang zur Physik auf der Nanometerskala.
- **25.03.96:** M. Bode, Regensburg, Germany (DPG Tagung): Korrelation nanostruktureller und lokaler elektronischer Eigenschaften von Fe/W(110) mittels RTS.
- 27.03.96: A. Wadas, M. Dreyer, M. Löhndorf, R. Wiesendanger, Regensburg, Germany (DPG-Tagung): Magnetic force microscopy results obtained in UHV.
- 27.03.96: M. Löhndorf, A. Wadas, and R. Wiesendanger, Regensburg, Germany, (DPG-Tagung): Untersuchung von dünnen magnetischen Schichten und Multilagensystemen mittels magnetischer Kraftmikroskopie.
- 28.03.96: H. Bluhm, A. Wadas, L. Szczesniak, K.-P. Meyer, and R. Wiesendanger, Regensburg, (DPG-Frühjahrstagung): Kontrastmechanismen bei der Abbildung ferroelektrischer Domänen mit dem Rasterkraftmikroskop.
- 28.03.96: U.D. Schwarz, P. Köster, H. Hölscher, O. Zwörner, and R. Wiesendanger, Regensburg, Germany, (DPG-Frühjahrstagung): *Friction force microscopy with well-defined tips on layered materials* (German).
- **16.04.96** S. Heinze, Forschungszentrum Jülich, Germany: *Theoretische Untersuchungen zu STM-Experimenten mitdem FLAPW-Code*.
- 18.04.96: U.D. Schwarz, P. Köster, O. Zwörner, and R. Wiesendanger, Hamburg, Germany, (DI User Meeting): Friction force microscopy with well-defined tips on layered materials.

- **18.04.96:** M. Löhndorf, Hamburg, Germany, (DI User Meeting): Magnetische Kraftmikroskopie an dünnen Co-Schichten.
- **18.04.96:** H. Bluhm, A. Wadas, and R. Wiesendanger, Hamburg, Germany, (DI User Meeting): *Abbildung ferroelektrischer Kristalle mit dem Kraftmikroskop*.
- **24.04.96:** J. Asmussen, Forschungszentrum Jülich, Germany: *Rasterthermomikroskopie.*
- **08.05.96:** M. Bode, Omicron User Meeting, Taunusstein, Germany: Korrelation struktureller und lokaler elektronischer Eigen- schaften von Fe/W(110) mittels RTS.
- **13.05.96:** A. Wadas, University of Manchester, (CAMST topical meeting on Magnetic Force Microscopy): Resolution and sensitivity of MFM Experiments in air and in UHV.
- **20.05.96:** S. Heinze, MPI Halle, Germany (Magnetic Nanostructures Workshop): What can we learn about magnetism and electronic structure from STM.
- **29.05.96:** U.D. Schwarz, W. Allers, H. Bluhm, H. Hölscher, P. Köster, O. Zwörner, and R. Wiesendanger, University of Konstanz, Konstanz, Germany, (Seminar): *Experimental and theoretical aspects of point contact friction* (German).
- 13.06.96: R. Wiesendanger, MPI für Mikrostrukturphysik, Halle, Germany: Korrelation struktureller, elektronischer und magnetischer Eigen schaften auf der Nanometerskala mittels Rastersondenmikroskopie und -spektroskopie.
- 16.-28.06.96: U.D. Schwarz, O. Zwörner, P. Köster, and R. Wiesendanger, Sesimbra, Portugal (NATO ASI workshop on Micro-/Nanotribology and Its Applications): Friction force microscopy with well-defined tips at layered materials and carbon compounds.
- **18.07.96:** M. Bode, Forschungszentrum Jülich, Germany: Korrelation struktureller und lokaler elektronischer Eigenschaften von Fe/W(110) mittels RTS.
- 13.09.96: U.D. Schwarz, H. Bluhm, H. Hölscher, O. Zwörner, P. Köster, W. Allers, and R. Wiesendanger, Tokyo Institute of Technology, Tokyo, Japan, (Seminar): Experimental and theoretical aspects of point contact friction.
- 13.09.96: U.D. Schwarz, H. Bluhm, H. Hölscher, O. Zwörner, P. Köster, W. Allers, and R. Wiesendanger, Advanced Technology Institute, Tokyo, Japan, (Seminar): Experimental and theoretical aspects of point contact friction.
- **17.10.96:** M. Bode, Pennsylvania, USA, (AVS symposium): Korrelation struktureller und lokaler elektronischer Eigenschaften von Fe/W(110) mittels RTS.

- 06.-07.11. 96: U.D.Schwarz, H. Bluhm, H. Hölscher, O. Zwörner, P. Köster, and R. Wiesendanger, Bad Honnef, Germany, (171th Seminar of the WE-Heraeus Foundation on Nanosciences and -technology): Nanotribology (German).
- **15.11.96:** A. Born, München, (Workshop zum Einsatz von Rastersondenverfahren zur Mikrocharakterisierung von Halbleiterbauelementen): Rastersondenmethoden zur Kapazitäts- und Magnetfeldmessung.
- 25.11.96: U.D. Schwarz, H. Bluhm, H. Hölscher, O. Zwörner, P. Köster, W. Allers, A. Schwarz, and R. Wiesendanger, University of Tübingen, Germany, (Seminar): *Fundamental investigations on point contact friction* (German).
- **29.11.96:** A. Born, Münster, (BMBF-Projekttreffen): Aufbau eines Prototyps zum Erzeugen und Auslesen von Nanometer-Strukturen im MHz-Bereich.
- **06.01.97:** R. Pascal, Ch. Zarnitz, M. Bode and R. Wiesendanger, Dreikönigstreffen, Bad Honnef, Germany, 168. WE-Heraeus-Seminar: Untersuchung der lokalen elektronischen Struktur von Gd auf W(110).
- **07.01.97:** S. Heinze, Dreikönigstreffen, Bad Honnef, Germany, 168. WE-Heraeus-Seminar: Zusammenhang zwischen elektronischer Struktur und Rastertunneltopographie (STM) und -Spektroskopie (STS) von Fe/W(110).
- 13.01.97: R. Dombrowski, Chr. Wittneven and R. Wiesendanger, 1. Workshop des Graduiertenkollegs Physik nanostrukturierter Festkörper, Timmendorfer Strand, Germany: Design eines UHV-Tieftemperatur-Rastertunnelmikroskops mit externem Vektormagnetfeld.
- **16.01.97:** R. Wiesendanger, University of Erlangen, Germany (SFB- Kolloquium): Rastersondenmikroskopie und Spektroskopie: Zugang zur Physik auf der Nanometerskala.
- **23.01.97:** R. Wiesendanger, University of Hamburg, Germany (Kolloquium): Rastersondenmikroskopie und Spektroskopie: Zugang zur Physik und Chemie auf der Nanometerskala.
- 18.02.97: A. Born, and R. Wiesendanger, DI User-Workshop 1997, Münster, Germany: Anwendungen des Rasterkapazitätsmikroskops für die Analyse von Integrierten Schaltkreisen.
- **20.02.97:** M. Kleiber, DI User Workshop, Münster, Germany: Untersuchung der magnetischen Eigenschaften von Kobalt-Nanopartikeln mit Magnetkraftmikroskopie.
- **07.03.97:** A. Schwarz, W. Allers, U. D. Schwarz, and R. Wiesendanger Johannes-Kepler-Universität, Linz, Austria: Construction of an ultrahigh vacuum lowtemperature scanning force microscope.

- 13.03.97: R. Wiesendanger, PTB Braunschweig, Germany (Seminar): Von der Nanostrukturierung mittels Rastersondenmethoden zur hochdichten Datenspeicherung: Probleme und Lösungsansätze.
- 17.03.97: H. Hölscher, U. D. Schwarz, and R. Wiesendanger, Spring Conference of the German Physical Society, Münster, Germany: A model for the lateral force microscope (German).
- 17.03.97: U. D. Schwarz, O. Zwörner, P. Köster, and R. Wiesendanger, Spring Conference of the German Physical Society, Münster, Germany: Comparative quantitative lateral force microscopy of carbon compounds in air and in argon atmosphere (German).
- 17.03.97: Chr. Wittneven, R. Dombrowski, S.H. Pan, and R. Wiesendanger, Frühjahrstagung der DPG, Münster, Germany: Aufbau eines UHV-TT-Rastertunnelmikroskops für Messungen im Vektormagnetfeld und erste Ergebnisse.
- 17.03.97: A. Born and R. Wiesendanger, Frühjahrstagung des Arbeitskreises Festkörperphysik bei der DPG, Münster, Germany: Analyse von Nanostrukturen integrierter Schaltungen mittels Rasterkapazitätsmikroskopie.
- **18.03.97:** A. Born. Münster, (DPG-Frühjahrstagung 1997): Aufbau eines Hochgeschwindigkeits-Rasterkapazitätsmikroskops.
- **19.03.97:** M. Kleiber, M. Winzer, N. Dix, and R. Wiesendanger, DPG Tagung, Münster, Germany: Herstellung von Nano-Dot-Arrays mittels Nanokugellithographie und Charakterisierung mittels Rastersensormethoden.
- **19.03.97:** M. Löhndorf, M. Kleiber, A. Wadas, and R. Wiesendanger, DPG Tagung, Münster, Germany: Untersuchung der mikromagnetischen Eigenschaften von Kobaltnanopartikeln mit Magnetkraftmikroskopie.
- 19.03.97: M. Dreyer, A. Wadas, M. Löhndorf, and R. Wiesendanger, DPG Tagung, Münster, Germany: Erzeugung magnetischer und topographischer Strukturen auf dünnen Kobaltfilmen mittels Atom-/Magnetkraftmikroskopie im UHV.
- **19.03.97:** A. Wadas, M. Dreyer, M. Löhndorf, and R. Wiesendanger, DPG Tagung, Münster, Germany: UHV magnetic force microscopy as a tool to study magnetomechanical effects on a nanometer scale.
- **19.03.97:** A. Born. Münster, (DPG-Frühjahrstagung 1997): Analyse von Nanostrukturen integrierter Schaltungen mittels Rasterkapazitätsmikroskopie.
- **20.03.97:** R. Pascal, Ch. Zarnitz, M. Bode, and R. Wiesendanger, Frühjahrstagung der DPG, Münster, Germany: Untersuchung der atomaren und lokalen elektronischen Struktur von Gd auf W(110).

- **20.03.97** W. Kresse, M. Bode, and R. Wiesendanger, Münster, Germany (DPG Tagung): UHV-Rastertunnelmikroskopie an p-Typ InAs.
- **20.03.97** Ch. Zarnitz, R. Pascal, M. Bode, and R. Wiesendanger, Münster, Germany (DPG Tagung): Untersuchung des Wachstums von Gd auf W(110) mittels RTM.
- **01.04.97:** W. Allers, A. Schwarz, U. D. Schwarz, and R. Wiesendanger, Lawrence Berkeley National Laboratories, Berkeley, USA, *Low temperature scanning force microscopy*.
- **01.04.97:** A. Wadas, M. Dreyer, M. Löhndorf, and R. Wiesendanger, INTERMAG Conference, New Orleans, USA: Novel writing using magnetic force microscopy in ultrahigh vacuum.
- **09.04.97:** R. Wiesendanger, FHI der MPG, Berlin, Germany: Rastersondenmikroskopie und -spektroskopie: Zugang zur Physik und Chemie auf der Nanometerskala.
- 13.05.97: A. Wadas, OMICRON Workshop, Taunusstein, Germany: Magnetic Force Microscopy in UHV.
- 14.05.97: R. Pascal, Ch. Zarnitz, M. Bode, M. Getzlaff, and R. Wiesendanger, Seminar der AG Kaindl, FU Berlin, Germany: Untersuchung des Gd(0001)-Oberflächenzustandes mittels Rastertunnelspektroskopie.
- **20.05.97:** R. Wiesendanger, University of Santa Barbara, USA (Seminar): Scanning Probe Methods in Science and Technology of Magnetism.
- **30.05.97:** M. Getzlaff, M. Bode, and R. Wiesendanger. Alesund, Norway (NCSS-4): *Electronic structure of hydrogen on Gd(0001)/W(110).*
- **27.06.97:** U. D. Schwarz, H. Bluhm, H. Hölscher, O. Zwörner, and R. Wiesendanger, University of Amsterdam, Amsterdam, The Netherlands, *Experimental and theoretical aspects of point contact friction*.
- **07.07.97:** A. Born, and R. Wiesendanger, Seminar "Molekulare Mustererkennung mit supramolekularen Strukturen", Tübingen, Germany: *Rasterkapazitätsmikroskopie an Halbleitern und Halbleiterbauelementen*.
- **11.07.97:** A. Born, Saarbrücken, (BMBF-Projekttreffen): Systemlösung für SXM-Speicherprojekte.
- **15.07.97:** A. Born, and R. Wiesendanger, Workshop '97 "Nanostructures at Surfaces and Interfaces", Konstanz, Germany: Nanostructure manipulation by magnetic force microscopy and scanning capacitance microscopy.
- 20.07.97: A. Born, 9th International Conference on Scanning Tunneling Microscopy/ Spectroscopy and Related Techniques, STM '97, Hamburg, Germany: Scanning capacitance microscope as a new storage device?!

- 21.07.97: H. Hölscher, O. Zwörner, and U. D. Schwarz, 9th International Conference on Scanning Tunneling Microscopy/ Spectroscopy and Related Techniques, STM '97, Hamburg, Germany: Consequences of atomic-scale "stick-slip" movement in scanning force microscopy.
- 21.07.97: M. Löhndorf, M. Kleiber, A. Wadas, D. Weiss, and R. Wiesendanger, 9th International Conference on Scanning Tunneling Microscopy/ Spectroscopy and Related Techniques, STM '97, Hamburg, Germany: Study of magnetic properties and magnetization switching of nanostructured Co-dot-arrays by magnetic force microscopy.
- 21.07.97: A. Wadas, M. Dreyer, M. Löhndorf, and R. Wiesendanger, 9th International Conference on Scanning Tunneling Microscopy/ Spectroscopy and Related Techniques, STM '97, Hamburg, Germany: Magnetostrictive writing by using magnetic force microscopy in UHV.
- 21.07.97: A. Schwarz, W. Allers, U. D. Schwarz, and R. Wiesendanger, 9th International Conference on Scanning Tunneling Microscopy/ Spectroscopy and Related Techniques, STM '97, Hamburg, Germany: Atomic resolution at 10 K in UHV with a new design of a scanning force microscope.
- 21.07.97: U. D. Schwarz, O. Zwörner, and P. Köster, 9th International Conference on Scanning Tunneling Microscopy/ Spectroscopy and Related Techniques, STM '97, Hamburg, Germany: Validity of contact mechanical models on the nanometer scale with respect to friction force spectroscopy measurements.
- 21.07.97: Chr. Wittneven, R. Dombrowski, S.H. Pan, and R. Wiesendanger, 9th International Conference on Scanning Tunneling Microscopy/ Spectroscopy and Related Techniques, STM '97, Hamburg, Germany: Development of a UHV Low Temperature STM in combination with a rotatable high magnetic field.
- **21.07.97:** A. Born, 9th International Conference on Scanning Tunneling Microscopy/ Spectroscopy and Related Techniques, STM '97, Hamburg, Germany: Analysis of electrical breakdown failures by means of SFM based methods.
- **21.07.97:** A. Born, and R. Wiesendanger, 9th International Conference on Scanning Tunneling Microscopy/ Spectroscopy and Related Techniques, STM '97, Hamburg, Germany: Scanning capacitance microscope as a tool for the characterization of integrated circuits.
- 21.07.97: R. Pascal, Ch. Zarnitz, M. Bode, and R. Wiesendanger, 9th International Conference on Scanning Tunneling Microscopy/ Spectroscopy and Related Techniques, STM '97, Hamburg, Germany: STS reveals domain-like behavior of the Gd(0001) surface state.

- 21.07.97: M. Bode, M. Getzlaff, S. Heinze and R. Wiesendanger, 9th International Conference on Scanning Tunneling Microscopy/ Spectroscopy and Related Techniques, STM '97, Hamburg Germany: Temperature Dependent Exchange Splitting of Gd(0001) Surface States studied by Scanning Tunneling Spectroscopy.
- **02.09.97:** A. Born, and R. Wiesendanger, Auftaktveranstaltung zur Hamburger Mikroelektronik Initiative, Hamburg, Germany: *Qualitätskontrolle von Halbleiterbauelementen mittels Rastersondenmethoden*.
- 28.10.97: M. Bode, R. Pascal, Ch. Zarnitz, M. Getzlaff, S. Heinze and R. Wiesendanger, Sandia Nat. Lab., USA: Scanning Tunneling Spectroscopy of thin metal films on W(110): correlation of structural and local electronic properties.
- **03.12.97:** Chr. Wittneven, R. Dombrowski, M. Morgenstern and R. Wiesendanger, Max-Planck-Institut, Stuttgart, Germany: Scanning Tunneling Spectroscopy on *n-InAs(110): Landau Level Quantization and Scattering of Electron Waves at* Dopant Atoms.
- **29.01.98:** R. Wiesendanger, University of Rostock, Germany (Kolloquium): *Tieftem*peratur-Tunnelspektroskopie an Halbleitern und magnetischen Systemen.
- **05.03.98:** A. Born, München, Germany (D.I. User-workshop): Grundlagen der Rasterkapazitätsmikroskopie.
- **17.03.98:** A. Born, München, Germany (Sonderseminar, Siemens AG): 2D-Dotierungsbestimmung von Halbleiterbauelementen mit Rastersondenmethoden.
- **23.03.98:** M. Bode, M. Getzlaff, and R. Wiesendanger, Regensburg, Germany (Frühjahrstagung der DPG): Rastertunnelspektroskopie von Gd(0001) mit Fe-Spitzen: Modellfall für spinpolarisiertes Tunneln.
- **24.03.98:** W. Allers, A. Schwarz, U. D. Schwarz, and R. Wiesendanger, Regensburg (Spring Conference of the German Physical Society): Atomic resolution at 10 K with a new design of a scanning force microscope (German).
- **24.03.98:** M. Getzlaff, M. Bode, S. Heinze, R. Pascal, and R. Wiesendanger, Regensburg, Germany (Frühjahrstagung der DPG): *Temperaturabhängige Austauschaufspaltung des Oberflächenzustandes von Gd(0001).*
- **24.03.98:** A. Born, Regensburg, (DPG-Frühjahrstagung 1998): Kapazitätsund Oberflächenpotential-Messungen an Halbleiterstrukturen mit Nanometer-Auflösung.
- **25.03.98:** H. Bluhm, U. D. Schwarz, and R. Wiesendanger, Regensburg (Spring Conference of the German Physical Society): Origin of the ferroelectric domain contrast observed in lateral force microscopy (German).

- **25.03.98:** Ch. Zarnitz, H. Tödter, H.J. Kim, M. Bode, M. Getzlaff, and R. Wiesendanger, Regensburg, Germany (Frühjahrstagung der DPG): *Präparation defektfreier Wolframkarbidoberflächen als Substrat zur Manipulation einzelner Atome bei 300 K.*
- 23.03.98 M. Dreyer, A. Wadas, M. Kleiber, and R. Wiesendanger, Regensburg, Germany (Frühjahrstagung der DPG): Untersuchung der schichtdickenabhängigen Domänenstruktur dünner Co-Filme auf Au(111) mittels Magnetkraftmikroskopie im Ultrahochvakuum.
- 25.03.98: R. Pascal, H. Tödter, M. Bode, M. Getzlaff, and R. Wiesendanger, Regensburg, Germany (Frühjahrstagung der DPG): RTM und RTS an Gd – Fe -Legierungen auf W(110).
- **25.03.98:** M. Getzlaff, M. Bode, R. Pascal, H. Tödter, and R. Wiesendanger, Regensburg, Germany (Frühjahrstagung der DPG): Adsorbate auf Gd(0001) eine kombinierte STM/UPS-Studie.
- 26.03.98 M. Morgenstern, R. Dombrowski, Chr. Wittneven und R. Wiesendanger, Regensburg, Germany (Frühjahrstagung der DPG): Interpretation spitzeninduzierter Zustände in der Rastertunnelspektroskopie von InAs(110)
- **26.03.98:** R. Pascal, H. Tödter, M. Bode, and R. Wiesendanger, Regensburg, Germany (Frühjahrstagung der DPG): *RTM und RTS an Tb/W(110)*.
- **26.03.98:** S. Heinze, S. Blügel, R. Pascal, M. Bode, and R. Wiesendanger, Regensburg, Germany (Frühjahrstagung der DPG): Zusammenhang zwischen Oberflächenzuständen und Antikorrugation in STM-Bildern von Übergangsmetalloberflächen.
- **08.05.98:** R. Wiesendanger, IFF, Forschungszentrum Jülich, Germany (Kolloquium): Vom Quantenpunkt zur quantenmagnetischen Speicherung: Tieftemperatur-Rastersondenspektroskopie an Halbleitern und Ferromagneten.
- 17.05.98 M. Morgenstern, M. Bode, M. Getzlaff, U. D. Schwarz, A. Wadas, and R. Wiesendanger, Hindsgavl, Danmark (Conference on Nanosciences for Nanotechnology): Nanocharacterisation by scanning probe microscopy.
- 16.06.98: R. Wiesendanger, Vancouver, Canada (3rd Int. Symp. on Metallic Multilayers, MML '98): Correlation between atomic, local electronic and magnetic structure of Gd(0001) thin films by variable-temperature scanning tunneling spectroscopy.
- 17.06.98: R. Wiesendanger, Vancouver, Canada (3rd Int. Symp. on Metallic Multilayers, MML '98): Thickness-dependent magnetic domain structures of ultrathin Co-films on Au(111) studied by magnetic force microscopy in ultrahigh vacuum.

- **03.07.98:** R. Wiesendanger, University of Bonn, Germany (SFB-Workshop): Rastersondenmikroskopie und -Spektroskopie an Festkrperdefekten.
- **06.07.98:** M. Getzlaff, M. Bode, and R. Wiesendanger, Copenhagen, Denmark (ICSFS-9): *Imaging of magnetic domains on the nanometer scale.*
- **06.07.98:** M. Morgenstern, München (Seminar der Gruppe Prof. Kotthaus, LMU München): Tieftemperatur-Rastertunnelspektroskopie an InAs(110): Landauquantisierung und die Streuung von Elektronenwellen an Dotieratomen.
- **08.07.98:** M. Getzlaff, M. Bode, and R. Wiesendanger, Copenhagen, Denmark (ICSFS-9): Vizualization of heterogeneously catalytic processes.
- **10.07.98:** M. Getzlaff, R. Pascal, M. Bode, and R. Wiesendanger, Copenhagen, Denmark (ICSFS-9): New insight into the surface magnetic properties of Gd(0001).
- **15.07.98:** R. Wiesendanger, University of Tübingen, Germany (Kepler-Vorlesung): Von der Mikroelektronik zur Nanoelektronik: Neue faszinierende Einblicke in das mikroskopische Verhalten von Festkrperelektronen.
- **21.07.98** W. Allers, U. D. Schwarz, A. Schwarz, and R. Wiesendanger, Osaka, Japan (First International Workshop on Noncontact Atomic Force Microscopy): Atomic resolution on van der Waals surfaces in non-contact mode with a newly designed scanning force microscope for operation in ultrahigh vacuum and at 10 K.
- 22.07.98 A. Schwarz, W. Allers, U. D. Schwarz, and R. Wiesendanger, Osaka, Japan (First International Workshop on Noncontact Atomic Force Microscopy): Dynamic scanning force microscopy on InAs(110): Simultaneous resolution of the As and In sublattice and observation of point defects.
- 23.07.98 H. Hölscher, A. Schwarz, U. D. Schwarz, W. Allers, and R. Wiesendanger, Osaka, Japan (First International Workshop on Noncontact Atomic Force Microscopy): Analysis of the frequency shift in non-contact atomic force microscopy.
- **01.09.98:** M. Bode, M. Getzlaff, and R. Wiesendanger, Birmingham, Great Britain (QSA-10): Vizualization of heterogeneously catalytic processes.
- **02.09.98:** M. Bode, R. Pascal, H. Tödter, M. Getzlaff, and R. Wiesendanger, Birmingham, Great Britain (ICSS-10): From Gd to ordered GdFe alloys.
- **02.09.98** A. Schwarz, W. Allers, U. D. Schwarz, and R. Wiesendanger, Birmingham, England (Fifth International Conference on Nanometer-scale Science and Technology (NANO V)): True atomic resolution on InAs(110) obtained by dynamic scanning force microscopy: Simultaneous Resolution of the As and In sublattice and observation of point defects.

- **02.09.98:** M. Getzlaff, M. Bode, and R. Wiesendanger, Cancun, Mexico (ICEM-14): *Vizualization of heterogeneously catalytic processes.*
- **03.09.98:** M. Bode, M. Getzlaff, and R. Wiesendanger, Birmingham, Great Britain (ICSS-10): Vizualization of the adsorption process of hydrogen on gadolinium.
- **03.09.98:** M. Bode, M. Getzlaff, and R. Wiesendanger, Birmingham, Great Britain (NANO-5): Scanning tunneling spectroscopy of Gd(0001) using Fe-coated tips: the model case of spin polarized tunneling.
- **04.09.98** H. Hölscher, A. Schwarz, U. D. Schwarz, W. Allers, and R. Wiesendanger, Birmingham, England (Fifth International Conference on Nanometer-scale Science and Technology (NANO V)): Towards the understanding of dynamic force microscopy: Analysis of the anharmonic cantilever oscillations.
- 15.09.98 H. Hölscher, W. Allers, U. D. Schwarz, A. Schwarz, and R. Wiesendanger, Basle, Switzerland (Third Conference on the Development and Technological Application of Nearfield Scanning Probe Methods): Determination of the tipsample potential from the frequency shift in the dynamic mode of atomic force microscopy.
- 15.09.98 A. Schwarz, W. Allers, U. D. Schwarz, and R. Wiesendanger, Basle, Switzerland (Third Conference on the Development and Technological Application of Nearfield Scanning Probe Methods): Dynamic scanning force microscopy of point defects on InAs(110)-(1×1) at low temperatures in UHV.
- 17.09.98 U. D. Schwarz, W. Allers, H. Hölscher, A. Schwarz, and R. Wiesendanger, Basle, Switzerland (Third Conference on the Development and Technological Application of Nearfield Scanning Probe Methods): *Towards the understanding* of dynamic scanning force microscopy: Analysis of the contrast mechanism.
- **16.10.98:** A. Born, München, (Werkstoffwoche 1998): Nanostrukturierung und Charakterisierung von magnetischen Materialien.
- 16.10.98 U. D. Schwarz, W. Allers, H. Hölscher, A. Schwarz und R. Wiesendanger, Lawrence Berkeley National Laboratory, Berkeley, USA: Dynamic scanning force microscopy at low temperatures in UHV.
- **20.10.98:** M. Getzlaff, M. Bode, and R. Wiesendanger, Mainz, Germany (Seminar): Abbildung magnetischer Domänen auf der nm-Skala mittels spin-sensitivem STM.
- **02.11.98** M. Bode, M. Getzlaff, and R. Wiesendanger, Baltimore, USA (AVS 98): *Visualization of heterogeneously catalytic processes.*

- 02.11.98 U. D. Schwarz, W. Allers, H. Bluhm, H. Hölscher, A. Schwarz, O. Zwörner, and R. Wiesendanger. The German Electron Synchrotron Facility (DESY), Hamburg: Forces and interactions at the atomic scale in scanning force microscopy (German).
- **04.11.98** M. Bode, M. Getzlaff, and R. Wiesendanger, Baltimore, USA (AVS 98): Imaging magnetic domains by spin-polarized scanning tunneling spectroscopy.
- **06.11.98:** R. Wiesendanger, University of Heidelberg, Germany (Kolloquium): Von der Mikroelektronik zur Nanoelektronik: Neue faszinierende Einblicke in das mikroskopische Verhalten von Festkörperelektronen.
- **06.11.98:** A. Pundt, M. Getzlaff, U. Laudahn, R. Wiesendanger, and R. Kirchheim, Clausthal-Zellerfeld, Germany (TMR-Meeting): *Imaging of hydrogen-induced plastic deformation of Gd films.*
- 10.11.98: M. Kleiber, Miami, USA (MMM '98): Thickness dependent magnetic domain structures of ultrathin Co/Au(111) films studied by means of magnetic force microscopy in ultrahigh vacuum.
- **12.11.98:** M. Kleiber, Miami, USA (MMM '98): Magnetization switching of submicrometer Co dots induced by a magnetic force microscope tip.
- 12.11.98: M. Bode, M. Getzlaff, S. Heinze, and R. Wiesendanger, Miami, USA (MMM-43): Temperature dependence of the exchange splitting of the Gd(0001) surface state: a scanning tunneling spectroscopy study.
- 14.12.98: R. Wiesendanger, HMI Berlin, Germany (Kolloquium): Von der Mikroelektronik zur Nanoelektronik: Neue faszinierende Einblicke in das mikroskopische Verhalten von Festkrperelektronen.
- **15.12.98:** R. Wiesendanger, TU Dresden, Germany (Kolloquium): *Tieftemperatur-Rastersondenspektroskopie: Zugang zur Physik auf der Nanometerskala.*

Poster

- **27.03.96:** M. Seider, U.D. Schwarz, and R. Wiesendanger, Regensburg, Germany, (Spring Conference of the German Physical Society): Loading-force dependent studies of topography and friction of particle traces in mica by means of scanning probe methods (German).
- 27.03.96: J. Asmussen and R. Wiesendanger, Regensburg, Germany, (DPG-Frühjahrstagung): Realisierung eines Rasterthermomikroskops und Anwendungen auf integrierte Schaltungen.

- 27.03.96: A. Born, C. Wobbe, and R. Wiesendanger, Regensburg, Germany, (DPG-Frühjahrstagung): Strukturierung und Analyse von Nanostrukturen mittels eines Rasterkapazitätsmikroskops.
- 27.03.96: C.Y. Hahn, T. Matsuyama, U. Merkt, and R. Wiesendanger, Regensburg, Germany, (DPG-Frühjahrstagung): Nanostrukturierung und Charakterisierung von Mikrobrücken aus Niob.
- **28.03.96:** H. Hölscher, U.D. Schwarz, and R. Wiesendanger, Regensburg, Germany, (Spring Conference of the German Physical Society): Simulation of the measuring process in friction force microscopy (German).
- 10.04.96: A. Wadas, M. Dreyer, M. Löhndorf, R. Wiesendanger, Seattle, USA, (IN-TERMAG'96): Magnetic force microscopy in UHV.
- 16.-28.06.96: H. Hölscher, U.D. Schwarz, and R. Wiesendanger, Sesimbra, Portugal (NATO ASI workshop on Micro-/Nanotribology and Its Applications): Simulation of atomic-scale processes in friction force microscopy.
- 04.-06.11.96: H. Hölscher, U.D. Schwarz, and R. Wiesendanger, Bad Honnef, Germany, (167th Seminar of the WE-Heraeus Foundation on Non-linear Physics of Complex Systems): Modelling of the scan process in friction force microscopy (German).
- 06.01.97: M. Getzlaff, J. Bansmann, L. Lu, and K.H. Meiwes-Broer, Dreikönigstreffen, Bad Honnef, Germany, 168. WE-Heraeus-Seminar: Kobaltinseln auf W(110): Elektronische und magnetische Eigenschaften.
- 17.03.97: D. Haude, U. Mick, Ch. Witt, M. Bode, and R. Wiesendanger, Frühjahrstagung der DPG, Münster, Germany: A UHV Scanning Tunneling Microscope for in-situ Studies of Thin Film Growth.
- 17.03.97: H. Hölscher, U. D. Schwarz, and R. Wiesendanger, Spring Conference of the German Physical Society, Münster, Germany: Consequences of the "stick-slip" movement in scanning force microscopy (German).
- 17.03.97: O. Zwörner, U. D. Schwarz, H. Hölscher, and R. Wiesendanger, Spring Conference of the German Physical Society, Münster, Germany: *Force spectro*scopical measurements of the velocity dependence of lateral forces (German).
- **19.03.97:** A. Born, and R. Wiesendanger, Frühjahrstagung des Arbeitskreises Festkörperphysik bei der DPG, Münster, Germany: *Aufbau eines Hochgeschwindigkeits-Rasterkapazitätsmikroskops*.
- 19.03.97: D. Haude, Chr. Witt, M. Bode und R. Wiesendanger, DPG Frühjahrstagung 97, Münster, Germany: Aufbau eines UHV-Rastertunnelmikroskops zur Untersuchung der Wachstumsstadien dünner Schichten.

- 19.03.97: W. Kresse, M. Bode und R. Wiesendanger, DPG Frühjahrstagung 97, Münster, Germany: UHV-Rastertunnelmikroskopie und -spektroskopie auf p-InAs(110).
- 19.03.97: R. Pascal, Chr. Zarnitz, M. Bode und R. Wiesendanger, DPG Frühjahrstagung 97, Münster, Germany: Untersuchung der atomaren und lokalen elektronischen Struktur von Gd/W(110) mittels STM und STS.
- 19.03.97: R. Pascal, Chr. Zarnitz, M. Bode und R. Wiesendanger, DPG Frühjahrstagung 97, Münster, Germany: Untersuchung des Wachstums von Gadolinium auf W(110) mittels Rastertunnelmikroskopie.
- **07.04.97:** H. Hölscher, U. D. Schwarz, and R. Wiesendanger, European Workshop on Microtechnology and Scanning Probe Microscopy, Mainz, Germany : Simulation of a scanned tip im scanning force microscopy.
- **07.04.97:** U. D. Schwarz, O. Zwörner, P. Köster, and R. Wiesendanger, European Workshop on Microtechnology and Scanning Probe Microscopy, Mainz, Germany: Quantitatively reproducible measurement of frictional forces in scanning force microscopy using well-defined tips.
- **30.05.97:** M. Getzlaff, M. Bode, and R. Wiesendanger, Alesund, Norway (NCSS-4): *Electronic structure of hydrogen on Gd(0001)/W(110).*
- 21.07.97: A. Born, A. Olbrich, M. Maywald, and R. Wiesendanger, 9th International Conference on Scanning Tunneling Microscopy/ Spectroscopy and Related Techniques, STM '97, Hamburg, Germany: Analysis of electrical breakdown failures by means of SFM based methods.
- 21.07.97: M. Dreyer, A. Wadas, M. Löhndorf, and R. Wiesendanger, 9th International Conference on Scanning Tunneling Microscopy/ Spectroscopy and Related Techniques, STM '97, Hamburg, Germany: Development of UHV magnetic force microscopy and its application to studies of domain structures of ultrathin Co films.
- 21.07.97 D. Haude, U. Mick, Ch. Witt, M. Bode, and R. Wiesendanger, 9th International Conference on Scanning Tunneling Microscopy/ Spectroscopy and Related Techniques, STM '97, Hamburg, Germany: A UHV Scanning Tunneling Microscope for in-situ Studies of Thin Film Growth.
- 21.07.97: H. Hölscher, O. Zwörner, and U. D. Schwarz, 9th International Conference on Scanning Tunneling Microscopy/ Spectroscopy and Related Techniques, STM '97, Hamburg, Germany: The velocity dependence of frictional forces in scanning force microscopy.

- 21.07.97: O. Zwörner, U. D. Schwarz, and R. Wiesendanger, 9th International Conference on Scanning Tunneling Microscopy/ Spectroscopy and Related Techniques, STM '97, Hamburg, Germany: Quantitative analysis of the frictional properties of carbon compounds eratures using friction force spectroscopy.
- 21.07.97: H. Tödter, R. Pascal, M. Bode, M. Getzlaff, and R. Wiesendanger, 9th International Conference on Scanning Tunneling Microscopy/ Spectroscopy and Related Techniques, STM '97, Hamburg, Germany: Scanning Tunneling Spectroscopy of Tb(0001)/W(110).
- 21.07.97: D. Haude, Ch. Witt, U. Mick, M. Bode, and R. Wiesendanger, 9th International Conference on Scanning Tunneling Microscopy/ Spectroscopy and Related Techniques, STM '97, Hamburg, Germany: A UHV scanning tunneling microscope for in-situ studies of thin film growth.
- 22.07.97: C. Zarnitz, R. Pascal, M. Bode, and R. Wiesendanger, 9th International Conference on Scanning Tunneling Microscopy/ Spectroscopy and Related Techniques, STM '97, Hamburg, Germany: Atomic and local electronic structure of Gd/W(110) in the submonolayer regime.
- 22.07.97: M. Bode, R. Pascal, S. Heinze, S. Blügel, and R. Wiesendanger, 9th International Conference on Scanning Tunneling Microscopy/ Spectroscopy and Related Techniques, STM '97, Hamburg, Germany: Nanostructural and local electronic properties of Fe/W(110): Scanning Tunneling Spectroscopy and abinitio calculations.
- 16.09.97: A. Born, C. Wobbe, and R. Wiesendanger, Forschungsforum '97, Leipzig, Germany: Hochgeschwindigkeits-Rastersondenmikroskop für Massenspeicheranwendungen.
- 17.09.97: J. Bansmann, L. Lu, and M. Getzlaff, ECOSS97, Enschede/NL: Magnetic Dichroism in Photoemission from oxygen adsorbed on iron and cobalt surfaces.
- 17.09.97: J. Bansmann, M. Getzlaff, G. Schönhense, M. Fluchtmann, and J. Brau, ECOSS97, Enschede/NL: Valence Band Photoemission from thin Fe Films: A Comparison of Experimental and Theoretical Results.
- **17.09.97:** M. Getzlaff, M. Bode, and R. Wiesendanger, ECOSS97, Enschede/NL: *Electronic structure of hydrogen on Gd(0001)/W(110).*
- **18.09.97:** M. Getzlaff, M. Bode, and R. Wiesendanger, Enschede, The Netherlands (ECOSS-17): *Electronic structure of hydrogen on Gd(0001)/W(110).*
- **21.10.97:** M. Bode, M Getzlaff, S. Heinze and R. Wiesendanger, AVS97, San Jose, USA: Variable Temperature Scanning Tunneling Spectroscopy of Gd(0001) Surface States.

- 17.03.98: M. Getzlaff, R. Pascal, H. Tödter, M. Bode, and R. Wiesendanger, Wandlitz, Germany (Workshop "Magnetism and electronic correlations in localmoment systems: Rare earth elements and compounds"): From Gd to ordered GdFe alloys – structure and electronic properties –.
- 23.03.98: M. Hennefarth, D. Haude, M. Bode, M. Getzlaff, and R. Wiesendanger, Regensburg, Germany (Frühjahrstagung der DPG): Wachstum und elektronische Struktur ultradünner Mn-Filme auf W(110).
- 24.03.98 K. Brooks, G. Meier, D. Grundler, M. Kleiber, D. Heitmann, and R. Wiesendanger, Regensburg, Germany (Frühjahrstagung der DPG): Herstellung und Charakterisierung von metallischen Dots mit quantisierter z-Magnetisierung.
- 24.03.98 J. H. Müller, U. D. Schwarz, and R. Wiesendanger, Regensburg, Germany (Frühjahrstagung der DPG): Design eines temperaturvariablen Rasterkraftmikroskops zur Charakterisierung organischer Proben.
- **26.03.98** U. D. Schwarz, H. Hölscher, W. Raberg, A. Hasbach, R. Wiesendanger, and K. Wandelt, Regensburg, Germany (Frühjahrstagung der DPG): *Analyse* des Abbildungsprozesses in der Kontaktkraftmikroskopie auf atomarer Skala.
- **07.07.98:** M. Getzlaff, R. Pascal, M. Bode, and R. Wiesendanger, Copenhagen, Denmark (ICSFS-9): *The adsorption process of hydrogen on Gd(0001).*
- **07.07.98:** M. Getzlaff, M. Hennefarth, M. Bode, and R. Wiesendanger, Copenhagen, Denmark (ICSFS-9): Ultrathin Mn films on W(110).
- **09.07.98:** M. Getzlaff, R. Pascal, H. Tödter, M. Bode, and R. Wiesendanger, Copenhagen, Denmark (ICSFS-9): *Characterization of ordered GdFe alloys.*
- **02.09.98** M. Bode, M. Hennefarth, G. Getzlaff, and R. Wiesendanger, Birmingham, Great Britain (ICSS-10): Growth of ultrathin Mn films on W(110).
- **03.09.98:** M. Getzlaff, M. Bode, and R. Wiesendanger, Cancun, Mexico (ICEM-14): From Gd to ordered GdFe alloys.
- **03.09.98:** M. Getzlaff, M. Bode, and R. Wiesendanger, Cancun, Mexico (ICEM-14): Adsorption of hydrogen on structured gadolinium.
- **05.11.98** M. Bode, M. Hennefarth, G. Getzlaff, and R. Wiesendanger, Baltimore, USA (AVS 98): Growth of ultrathin Mn films on W(110).
- 09.11.98 H. Hölscher, U. D. Schwarz, A. Schwarz, W. Allers, O. Zwörner und R. Wiesendanger, Marburg (186th WE-Heraeus-Seminar on "Statistical Mechanics, Nonlinear Dynamics and Turbulence"): Analysis of nonlinear dynamics in scanning force microscopy.

Chapter 8 Talks Given by Guests

- **26.01.96:** Prof. Dr. H. Thomas (Universität Basel): Mechanismus der atomaren Reibung.
- **01.02.96:** Prof. Dr. J. Kirschner (MPI für Mikrostrukturphysik Halle): Korrelation von magnetischen und strukturellen Eigenschaften von dünnen Schichten.
- **05.02.96:** Dr. J. Wintterlin (FHI Berlin): STM-Untersuchungen zu Mechanismen und Dynamik chemischer Oberflächenprozesse.
- **08.02.96:** Dr. G. Meyer (FU Berlin): Tieftemperatur-Rastertunnelmikroskopie von CO auf Cu(211) - Manipulation von Atomen und Molekülen.
- **22.04.96:** Prof. Dr. Klaus von Klitzing (MPI für Festkörperforschung, Stuttgart): Elektische Eigenschaften niedrigdimensionaler Systeme.
- **22.04.96:** Dr. Leo Kouwenhoven (Technical University of Delft, The Netherlands): *Electron Transport through Quantum Dots.*
- **22.04.96:** Prof. Dr. Michel H. Devoret (CEA-Saclay, France): Single electron Transfer in Metallic and Superconducting Nanostructures.
- **22.04.96:** Prof. Dr. Akira Tonomura (Hitachi, Saitama, Japan): Real-time Observation of the Dynamics of Vortices in Superconductors using Electron Waves.
- **22.04.96:** Dr. Ali Yazdani (IBM Almaden Research, San Jose, USA): Atomic scale Perspectives of Superconductivity and Magnetism: New View from a Cold STM.
- **29.04.96:** Dipl.-Phys. K. Schiffmann, (Fraunhofer Institut, Braunschweig): Struktur- und Reibungsuntersuchungen an metallhaltigen amorphen Kohlenwasserstoff-Nanocompositen mittels Rastersondenmikroskopie.
- **02.05.96:** Prof. Dr. G. Ertl (FHI der MPG, Berlin): Selbstorganisation bei Oberflächenreaktionen.
- **27.06.96:** Prof. Dr. H. Kurz (RWTH Aachen): Ultrakurz und ultraklein: Neue Dimensionen in der Informationstechnik.

- **27.09.96:** Dr. M. Morgenstern (Forschungszentrum Jülich): Die Eisbilage auf Pt(111): Nukleation, Struktur und Schmelzverhalten.
- **01.10.96:** Dr. M. Hohage (Forschungszentrum Jülich): Kinetik der Pt(111)-Oberfläche bei tiefen Temperaturen.
- **04.11.96:** Dr. H.W.P. Koops (Telekom Forschungszentrum Darmstadt): Additive dreidimensionale Elektronenstrahl-Nanolithographie und ihre Anwendungen für die Mikrotechnik.
- **18.11.96:** Dr. M. Ziese (Universität Leipzig): Struktur des Flußliniengitters in supraleitenden Filmen und Schichtsupraleitern.
- **25.11.96:** Dr. K. Nehrke (Universität Pavia): NMR-Untersuchung der magnetischen Tieftemperaturphase in PrBa₂Cu₃O₆+x.
- **12.12.96:** Dr. M. Wendel (LMU München): Herstellung und Untersuchung von Halbleiterbauelementen mit dem Rasterkraftmikroskop.
- **13.01.97:** Dipl.Phys. K. Schiffmann (Fraunhofer-Institut f. Schicht- u. Oberflächentechnik, Braunschweig): *Mikroverschleißmessungen an Metall-C:H-Schichten*.
- **20.01.97:** Dr. Ph. Ebert (Forschungszentrum Jülich): Untersuchungen von Defekten auf III-V-Halbleiter-Oberflächen.
- **21.04.97:** Dipl.-Phys. G. Marx (Universität Mainz): Anwendung eines Photoemmissionselektronenmikroskops in der Oberflächenphysik.
- **15.05.97:** Prof. Dr. H. Zabel (Ruhr-Univ. Bochum): Magnetismus und Supraleitung in metallischen Schichtsystemen.
- **23.06.97:** Prof. Dr. K. L. Johnson (University of Cambridge): Continuum mechanics modelling of adhesion and friction in a single asperity contact.
- **08.07.97:** Prof. Dr. O. Marti (Universität Ulm): Statische und dynamische Polymereigenschaften auf der Nanometerskala.
- **07.08.97:** Dr. H. Bluhm (Berkeley, USA): Wachstums- und Reibungseigenschaften von dünnen Eisfilmen.
- **28.10.97:** Prof. Dr. A. Heuberger (Fraunhofer Institut für Siliziumtechnologie, Itzehoe): Industrielle Bedeutung der Mikrosystemtechnik.
- **28.10.97:** Dr. E. Demm (Siemens AG, München): Analytik in der Halbleitertechnik im Mikro- und Nanostrukturbereich.
- **28.10.97:** Dr. H. Pohlmann (Philips GmbH, Hamburg): *Mikrostrukturierung* zur Herstellung von Halbleiterschaltungen.
- **28.10.97:** Dr. H. van den Berg (Siemens AG, Erlangen): Magnetische Mikround Nanostrukturen in neuartigen Bauelementen und Systemen.
- **28.10.97:** Dr. P. Schintag (Philips GmbH, Hamburg): Magnetoresistive Sensoren und Sensorsysteme.

- **28.10.97:** Dr. U. Hintze (Beiersdorf AG, Hamburg): Ultrastrukturanalyse eine effektive Technologie für die Entwicklung hautpflegender und klebender Produkte.
- 18.11.97: Prof. Dr. M. Horn-von Hoegen (Institut für Festkörperphysik, Universität Hannover): Manipulation von Oberflächen mit Surfactants: von perfekter Heteroepitaxie bis zu Quantendrähten.
- **19.11.97:** Dr. T. Fritz (Institut für Angewandte Photophysik, TU Dresden): Organisch-Anorganische Heteroepitaxie aromatischer Moleküle auf Au(111)-Oberflächen.
- **21.11.97:** Dr. U. Keil (Mikroelektronik-Zentrum, Lyngby, Dänemark): Rastertunnelmikroskopie mit Subpikosekunden-Zeitauflösung.
- 24.11.97: Dr. L. Chi (Physikal. Institut, Universität Münster): Herstellung von geordneten organischen Schichtsystemen und ihre Charakterisierung mittels Rastersondenmikroskopie.
- **25.11.97:** Dr. U. Steiner (Institut für Physik, Universität Konstanz): Polymere an Oberflächen und Grenzflächen: Phasenverhalten, Benetzung und Strukturbildung.
- **26.11.97:** Dr. J. Wrachtrup (Institut für Physik, TU Chemnitz): Einzelmolekülspektroskopie: Von der Beobachtung isolierter atomarer Defekte zur Spektroskopie an einzelnen biomolekularen Komplexen auf Oberflächen.
- **28.11.97:** Dr. H. P. Oepen (MPI für Mikrostrukturphysik, Halle): Magnetische Mikrostrukturen und Anisotropien in ultradünnen Ferromagneten.
- **02.12.97:** Dr. I. Michely (IGV, Forschungszentrum Jülich): Atomare Prozesse beim Schichtwachstum - Molekularstrahlepitaxie, ionenstrahlgestützte Deposition und Sputterdeposition im Vergleich.
- **03.12.97:** Dr. F. U. Hillebrecht (Institut f. Angewandte Physik, Universität Düsseldorf): *Magnetische Filme und Nanostrukturen*.
- 16.12.97: Dr. R. Houbertz-Krauß (Institut für Experimentalphysik, Universität Saarbrücken): Untersuchungen zum elektronischen Transport an niedrigdimensionalen Heterosystemen.
- **26.01.98:** Dr. Ch. Renner (Universität Genf): New evidences from scanning tunneling spectroscopy for a non-BCS behaviour in high temperature super-conductors.
- **02.02.98:** Dr. M. Donath (MPI für Plasmaphysik, Garching): Magnetische Oberflächenzustände: Untersuchungen mit spinaufgelöster inverser Photoemission.
- **04.05.98:** Dr. M. Wassermeier (Paul-Drude-Institut Berlin): Rastertunnelmikroskopie der Rekonstruktionen von GaAs- und GaN-Oberflächen.

- 14.05.98: Dipl.-Phys. M. Adameck (TU Hamburg-Harburg): Hochauflösende Rastermethoden zur Charakterisierung von NLO-Polymerfilmen.
- **18.05.98:** Dr. W. Raberg (Universität Bonn): Rastersondenmikroskopie an amorphen Oberflächen.
- **10.06.98:** Prof. Dr. R. Wäppling (Uppsala University): Relevant research for magnetic storage.
- **15.06.98:** Dipl.-Phys. Waldemar Babik (Universität Bonn): Untersuchung der Hyperfeinwechselwirkung in verschiedenen magnetischen Matrizen mit der Methode der Kernorientierung sowie statischer und dynamischer NMR/ON.
- **29.06.98:** Dipl.-Phys. B. Gotsmann (Uni Münster): Rekonstruktion von Spitze-Probe-Wechselwirkungskräften aus der dynamischen Kraftspektroskopie.
- **21.07.98:** Dipl.-Phys. A. Wachowiak (Uni Heidelberg): Rastertunnelmikroskopie bei tiefen Temperaturen.
- **16.09.98:** Dipl.-Phys. A. Knoll (Uni Würzburg): Aufbau und Charakterisierung eines UHV-Rastertunnelmikroskops.
- **08.10.98:** Prof. Dr. S. A. Nepijko (Ukrainian Natl. Acad. of Sciences, Kiev): Magnetische Eigenschaften von Clustern: Einsatz der Interferenzelektronen-Mikroskopie für die Messung magnetischer Eigenschaften.
- **29.10.98:** Prof. Dr. H. van Kempen (University of Nijmegen): Scanning probe spectroscopy of III-V semiconductors.
- **29.10.98:** Prof. Dr. K.-H. Rieder (FU Berlin): Atomic manipulations with a low-temperature scanning tunneling microscope.
- **30.10.98:** Dr. J. Weis (MPI für Festkörperforschung, Stuttgart: Single electrons as a probe for mesoscopic systems.
- **30.11.98:** Dr. M. Reichling (FU Berlin): *Hochaufgelöste Rasterkraftmikroskopie* an Fluoridoberflächen.
- **07.12.98:** Dr. M. Schmid (TU Wien): Rastertunnelmikroskopie an Metalloberflächen: Legierungen und Schichtwachstum.
- 14.12.98: Dr. A. Olbrich (Siemens, München): AFM-Messungen mit leitfähiger Spitze zur Charakterisierung dünner Dielektrika.

Chapter 9 Lectures and Courses at the University of Hamburg

Einführung in die Struktur der Materie (Kursusvorlesung)

Aktuelle Themen der Festkörperphysik

Einführung in die Nanotechnologie

Einführung in die Rastertunnel- und Rastersensormikroskopie I und II

Anwendungen der Rastertunnel- und Rastersensormikroskopie I und II

Oberflächen- und Grenzflächenphysik mit spinpolarisierten Elektronen

Ringvorlesung "Physik der Mikrostrukturen"

Seminar über hochauflösende Mikroskopie an Festkörperoberflächen

Seminar über lokale Methoden in der Festkörperphysik

Seminar über Nahfeldgrenzflächenphysik

Proseminar über Rastertunnelmikroskopie und verwandte Rastersensormethoden

Proseminar über Neue Materialien

Übungen zur Einführung in die Struktur der Materie

Übungen zur Einführung in die Rastertunnel- und Rastersensormikroskopie

Laborpraktikum

Diplompraktikum

Chapter 10

Contributions to International Organisations and Journals

10.1 Organisations

- Since 1996: Chair of the Nano-Science and Technology Division of the German Vacuum Society (DVG)
- Since 1998: Chair of the Nano Science Division of the International Union for Vacuum Science, Techniques and Applications (IUVSTA)
- Co-Organiser of the "2nd International Symposium on the Physics of Micro- and Nanostructures" (Hamburg 1996)
- Member of the International Program Committee of the "4th International Conference on Nanometer-Scale Science and Technology (NANO-IV)" (Beijing, China 1996)
- Member of the Local Organising Committee of the "13th International Symposium on the Reactivity of Solids" (Hamburg 1996)
- Co-Chairman of the "1st International Symposium on Scanning Probe Spectroscopy (SPS '97)" (Poznan, Poland 1997)
- Chairman of the "9th International Conference on Scanning Tunneling Microscopy and Related Techniques (STM '97)" (Hamburg 1997)
- Co-Organiser of the "3rd International Symposium on the Physics of Micro- and Nanostructures" (Hamburg 1997)
- Organiser of the Symposium on "Scanning Probe Microscopy and Spectroscopy at low Temperatures" at the Spring Meeting of the German Physical Society (Regensburg 1998)

- Member of the Organising Committee of the "1st International Workshop on Non-Contact Atomic Force Microscopy" (Osaka, Japan 1998)
- Co-Organiser of the "2nd Workshop on Quantum Materials" (Hamburg 1998)

10.2 Journals

- Co-Editor of Applied Physics A (Topic: Nanostructure Physics and Technology)
- Co-Editor of the Springer Series in Nano-Science and Technology

Chapter 11

How to reach us

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