Druck und Einband: Printy Digitaldruck, München (http://www.printy.de)
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Preface

Dear friends, partners, sponsors, and alumni of the Walter Schottky Institut!

Once again we are pleased to present to you our annual report, to keep you updated about the diverse scientific activities at the WSI in the year 2009. As usual, the main scientific results and the status of ongoing research projects are summarized in the form of two-page abstracts, complemented by a listing of publications, collaborations, lectures, as well as diploma and PhD theses.

During 2009, almost 90 PhD students as well as more than 60 Diploma and Master students worked in the crowded laboratories and offices of the Institute. 15 PhD theses, 26 Diploma theses, and five Master theses were successfully completed. All PhD theses were published in the Institute’s book series “Selected Topics of Semiconductor Physics and Technology”, which now contains 114 volumes. We are particularly proud of the fact that Werner Hofmann and Sabrina Niesar were honored with the E.ON Future Award 2009 for the best PhD and Diploma Thesis, respectively. Other awards were bestowed upon Andreas Reitinger (Best Oral Presentation Award, Diamond 2009) and Ian Sharp (Carl von Linde Junior Fellowship, TUM Institute for Advanced Study). In addition, more than 120 publications in international journals and numerous invited and contributed talks at conferences all over the world have added to the scientific visibility of the WSI. I would like to express my sincere gratitude to all institute members who have contributed to this, both, directly or indirectly and, in particular to all funding organizations for their financial support!

I also have the sad duty to mention the unexpected death of one of our PhD students, Kaveh Kashani-Shirazi, who died in a tragic accident on June 13, 2006. Kaveh was not only an excellent scientist but also a very lively and active member of the WSI and liked by all of those who knew him. Kaveh will be missed but not forgotten!

Garching, June 2010

Martin Stutzmann
Kaveh Kashani-Shirazi

Our PhD’s of 2009

Felix Hofbauer

Werner Hofmann

Robert Lechner

Nebile Isik

Johanna Rössler

Till Andlauer
1. **The Walter Schottky Institute**

**Brief History**

The Walter Schottky Institute (WSI) is a Central Institute of the Technical University of Munich (TUM). It was founded in order to strengthen the interaction between basic physics and semiconductor electronics research and development. After the decision was made to create such an interdisciplinary research institute in early 1986 it took about two years until the new laboratories became operational in May 1988. This exceptionally short time for planning and construction of a modern institute building was made possible by the excellent cooperation between the Siemens AG, the Bavarian ministries, and TUM.

**Resources**

The WSI building contains laboratories and offices with a total area of about 2400 m². It is well equipped with state-of-the-art facilities for semiconductor preparation, characterization, and device technology. The main resources are listed in the following:

**Materials preparation and semiconductor technology**

- Epitaxy systems for GaAs, InP, GaSb, GaN/AlGaN, and SiGe based heterostructures (MBE, CBE, plasma-induced MBE, MOCVD)
- Ultrahigh purity GaAs MBE machine (electron mobilities > 10 Mio cm²/Vs)
- UHV evaporation equipment for SiGe on glass
- Plasma-enhanced Si-CVD
- Pulsed laser processing
- Laboratory for surface modification / biofunctionalization
- 250 m² class 100 clean room facility with photolithography, e-beam lithography, reactive ion etching, metallization

**Characterization and spectroscopy tools**

- High resolution X-ray diffraction
- Atomic force microscopy
- Electron microscopy and EDX equipment
- Photoluminescence and Raman spectroscopy (from IR to UV)
- DLTS, Optical DLTS, CV-profiling
- FTIR spectroscopy
- X-ray photoelectron spectroscopy
- High frequency parameter analyzers
- Electron spin resonance (ODMR, EDMR)
- Low temperature high magnetic field setups for magneto-transport and magneto-optics
- Special characterization facilities for laser diodes

**Computational facilities**

- High end workstations
Research groups

The number of researchers at the Walter Schottky Institut has grown continuously. Today it accommodates the research groups headed by Gerhard Abstreiter, Markus-Christian Amann, Martin Brandt, Anna Fontcuberta i Morral, Jonathan J. Finley, Gregor Koblmüller, Alexander Holleitner, Ian Sharp, Martin Stutzmann, and Peter Vogl, with a total headcount of about 140, including junior research group leaders, scientific and technical staff, postdocs and visiting researchers, secretaries, and doctorate as well as diploma (master) students. Out of these, about 30 positions are funded by TUM, while basically all the doctorate positions are financed via research projects with external funding. Available laboratory and office space is by far not sufficient anymore and the WSI urgently needs an expansion. A new Center for Nanotechnology and Nanomaterials is currently in the planning stage and it is expected that this shared facility will become operational in fall 2010.

The main research interests are:

- Fabrication and characterization of new semiconductor materials, material combinations, as well as functionalisation of surfaces
- Development of novel methods for fabrication and characterization of nanostructures
- Basic physics with emphasis on electronic and optical properties of low dimensional systems
- Realisation of new semiconductor devices for applications in ultrafast electronics, optoelectronics, and as biological / chemical sensors
- Theory and simulation of modern semiconductor materials and devices

Experimental Semiconductor Physics I (Gerhard Abstreiter, Jonathan J. Finley, Gregor Koblmüller, Anna Fontcuberta i Morral, Alexander Holleitner):

Research projects of these groups deal with various aspects of electronic and optical properties of low-dimensional, mesoscopic semiconductor structures, the heteroepitaxy of group IV and III-V semiconductors, the development of novel methods for lateral patterning and self assembly of quantum wires and quantum dots, the use of various analytical tools for the characterization of nanometer-sized structures in collaboration with external groups, as well as the fabrication and test of new, unconventional electronic and optoelectronic devices. Examples for basic research are optical spectroscopy of single quantum dots, cleaved edge overgrowth on GaAs, magnetotransport in ultrahigh mobility GaAs heterostructures as well as electronic transport and tunneling in edge channels and one-dimensional systems. Device and technology oriented work aims at novel concepts for charge and spin storage in quantum dots, coherent devices based on quantum dots for future quantum information technology, photonic crystal microcavities for efficient single photon sources and the test of semiconductor nanostructures for chemical/biological sensors. A new area of research is the controlled manipulation of oligonucleotides on gold surface for possible protein detection and the development of SOI based lab-on-a-chip systems. Also of increasing interest are carbon based nanostructures and combinations with organic molecules.
Semiconductor Technology (Markus-Christian Amann):

The research activities in this group are concentrated on modern technologies for III-V compound semiconductors and their use for developing advanced electronic and optoelectronic devices. This comprises the development of epitaxial, patterning, microstructuring, etching and coating techniques as well as the design and fabrication of semiconductor laser diodes and other photonic components. The applied material systems are GaAs-AlGaAs, InGaAsP-InP and antimonite based compounds that are grown with molecular beam epitaxy (MBE) and chemical beam epitaxy (CBE) with an accuracy in the nanometer regime. Device structuring in the 100-nm-range is obtained by using electron-beam lithography. Reactive ion etching enables the well-defined processing of the various devices with a high material selectivity of the etching rate. The group is also well equipped with evaporation and sputtering techniques for passivation and contacting of the devices. Among the key devices are single-mode and wavelength-tunable laser diodes for the wavelength range between 1300 and 2200 nm using lattice-matched and strained InGaAsP layers on InP substrates, InGaAsP and AlInGaAs vertical cavity surface-emitting laser diodes in the 1300-2000 nm wavelength range. Recent work also covers the development of wavelength-tunable laser diodes for wavelengths above 2 µm using antimonite based compounds and quantum cascade lasers in the range of 5 to 15 µm for gas sensing. In the field of high-frequency electronic devices, the generation of rf-output power at millimeter-wave frequencies up to 300 GHz is aspired. Active transit-time diodes as well as passive varactor structures for multipliers are investigated.

Experimental Semiconductor Physics II (Martin Stutzmann, Martin Brandt):

The work of this semiconductor physics group deals with various aspects of new and non-conventional semiconductor materials and material combinations:

- semiconductors with a wide bandgap (GaN, InGaN, AlGaN, diamond, SiC),
- disordered semiconductors (amorphous, nanocrystalline, and polycrystalline),
- advanced thin film systems (silicon-based luminescent layers, thin film solar cells, organic/anorganic heterosystems, biofunctionalized semiconductors, semimagnetic semiconductors)

Most of these material systems are prepared by suitable deposition techniques (MBE, Plasma-enhanced CVD, e-beam evaporation, sputtering). Their efficient optimization is based on the large pool of structural, optical, and electrical characterization techniques available in our institute. Complementary to the usual spectroscopic techniques we have developed and employ a variety of highly sensitive methods which enable us to study in particular the influence of defects on the electronic performance of materials and devices. Such techniques include subgap absorption spectroscopy, optically induced capacitance spectroscopy and, in particular, modern spin resonance techniques which are applied to various materials systems and devices for spintronics.

In addition to the preparation and characterization of new semiconductor materials we also work on the modification and processing of semiconductors with pulsed high power laser
systems (laser-crystallization, holographic nanostructuring, laser-induced etching) and investigate the potential of new material systems for novel device structures. Recent examples include nanostructured thin film solar cells, high electron mobility transistors based on AlGaN/GaN heterostructures, as well as UV-detectors, sensors and biosensors.

**Theoretical Semiconductor Physics (Peter Vogl):**

The activity of this group focuses on the theoretical study of structural, electronic and optical properties of semiconductors. The work deals with mesoscopic structures, new materials, and novel devices, in the attempt to understand their basic physics and to predict their behavior. The research conducted in the group aims at the development of theoretical tools which can support present experimental activities and propose new ideas and solutions for the future, as indicated by the strong interaction with industrial laboratories and with engineering university departments. Sophisticated quantum mechanical calculations are used for determining the electronic structures and the optical properties of a variety of semiconductor materials and systems. The information from such fundamental studies constitutes the basis for the analysis of the transport properties of such materials and for the development of reliable numerical tools for device modeling. Recent activities in this group include ab-initio studies of spin devices, prediction of novel magnetic field effects, development of multiscale methods for film growth, and the design and modeling of semiconductor based quantum information devices.

The research activity of the institute thus covers a wide spectrum from basic physics in low-dimensional semiconductor structures to the development of novel or improved electronic, optoelectronic and sensor devices based on semiconductor hetero- and nanostructures. The close collaboration between the different groups and the availability of various experimental techniques are the essential basis for the successful development of novel semiconductor devices. Close contacts with industrial partners have also proven to be very fruitful and stimulating in developing new ideas and in following new directions which may be relevant for future applications.

Apart from the extensive research activities all groups are involved in teaching within their respective departments. Besides the usual teaching responsibilities in undergraduate and graduate courses, special emphasis is put on the education of diploma and doctoral students in the physics and technology of present and future nano-devices and of low dimensional semiconductor structures.
Fundamental Semiconductor Physics
Ultrafast transport current in GaAs based photoswitches

L. Prechtel, S. Manus, D. Schuh, W. Wegscheider, and A. Holleitner

Commercial sources and detectors for terahertz electromagnetic radiation (THz-EMR) have lead to numerous applications in materials and life sciences. Nevertheless, the physical processes involved in THz-EMR generation are still subject of fundamental research. For instance, THz-EMR is generated by a non-linear wave conversion of a short laser pulse in an electro-optic medium, by the photoelectric effect of a surface built-in electric field, and by photo-Dember mechanisms. Furthermore, hot carrier dynamics and collective electron-hole plasma processes can be probed by analyzing the emitted THz-EMR from a semiconductor test structure.

THz-EMR can also be generated in photoswitches (PS). These are semiconductor slabs which are voltage-biased across two surface metal electrodes. The PS is typically excited by an optical femtosecond pulse. In turn, a short current response in the PS leads to THz-EMR. The optical illumination profile determines which optoelectronic phenomenon dominates the current response within the PS. A spatially uniform illumination of the PS gives rise to a transport current pulse, which is limited by the lifetime of the optically excited charge carriers. Since sub-picosecond carrier lifetimes can be achieved in a variety of materials (e.g. GaAs, Si, InGaAs), this kind of THz-EMR generation is widely exploited in scientific and technological applications. A non-uniform illumination leads to a dominating displacement current pulse whose duration can be shorter than the carrier lifetime. While a theoretical description of both mechanisms has been developed, the actual real-space motion of the transport current pulse has not been experimentally characterized so far.

We reported on space- and time-resolved photocurrent studies of a PS.[1] The PS is part of a coplanar stripline circuit (Fig. 1). An optical probe-pulse is focused at a position \( x_0 \) in between the two CPS. Hereby, the two metal strips of the CPS form a PS for the stripline circuit (Fig. 1). After optical pump-excitation, an electro-magnetic pulse starts to propagate along the CPS. At a distance \( d \), the field-probe of the sampling circuit is short-circuited by an optical probe-pulse for the duration of the lifetime of the photo-generated charge carriers in the LT-GaAs at this position. The transient electric field of the CPS located at the field-probe during this time-period drives the current \( I_{\text{Sampling}} \) in the sampling circuit (Fig. 1). Most importantly, varying the time-delay \( \Delta t \) between the optical pump-pulse and the probe-pulse gives access to the time evolution of \( I_{\text{Sampling}} = I_{\text{Sampling}}(\Delta t) \).

Fig. 1: Optoelectronic on-chip detection scheme. An optical pump-pulse is focused between two metal strips forming a coplanar stripline at the position \( x_0 \). This gives rise to a photocurrent \( I_{\text{Photo}} \) in the stripline circuit. At a distance \( d \), the probe-pulse triggers the sampling circuit, which consists of a field probe, a current-voltage-amplifier, and a lock-in-amplifier.

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Supported by DFG-project HO 3324/4, the German excellence initiative via the "Nanosystems Initiative Munich" (NIM), and the “Center of NanoScience (CeNS)” in Munich.

Fig. 2: (a) Time-resolved $I_{\text{Sampling}}$ for several excitation positions $x_0$ with 5.6 $\mu$m $\leq x_0 \leq$ 9.4 $\mu$m in steps of 625 nm with a first (open triangle) and a second peak (filled triangle). Lines are fits. (b) Schematic band structure of GaAs along the $x_0$-direction at $V_{SD}$ = 3 V. The pump-laser excites charge carriers at position $x_0$. (c) Electric field $|\vec{E}|$ before (solid line) and directly after pump-laser excitation (dashed line). (d) Relative time-delay $t_{12}$ between the two peaks in Fig.2(a) as a function of $x_0$ for $E_{\text{Laser}}$ = 1.59 eV (open circles) and $E_{\text{Laser}}$ = 1.51 eV (full circles).

Fig. 2(a) depicts $I_{\text{Sampling}}$ as a function of $\Delta t$ for varying $x_0$. We observe two peaks of $I_{\text{Sampling}}$. For all $x_0$, we observe that the first peak occurs at the same $\Delta t$, and we interpret it to result from a displacement current in the PS. Fig. 2(b) shows the schematic of the bandstructure along $x_0$. The optical pump-pulse excites charge carriers at position $x_0$. Before illumination, the electric field $|\vec{E}| = V_{SD} / d$ is constant in between the two contacts. The excitation laser generates a high local electrical conductivity in the vicinity of $x_0$. Therefore, the electric field drops to zero at the position $x_0$ directly after laser excitation [dashed line in Fig. 2(c)], while it increases in the remaining area to sustain the applied bias $V_{SD}$. The resulting displacement current density is given by $j_D = \varepsilon_{\text{GaAs}} \varepsilon_0 \partial E / \partial t$. The second peak in Fig. 2(a) is found to shift in $\Delta t$ with respect to the first peak as a function of $x_0$. The difference of the center positions $t_{12}$ of the two peaks versus $x_0$ is shown in Figure 2(d). A linear fit allows us to deduce propagation velocities of $\sim 10^8$ cm/s. This value exceeds typical Fermi- and quantum velocities of single-particle charge excitations in GaAs at the utilized $E_{\text{Laser}}$. However, they are consistent with values for collective electron-hole plasma excitations within a GaAs PS. We therefore conclude that a collective excitation of an electron-hole plasma and not single charge-carriers dominates the transport current on ultrafast timescales in GaAs photoswitches.

Our results suggest that the presented ultrafast transport current phenomena may occur in further GaAs based optoelectronic circuits comprising e.g. nanowires[2,3] and quantum wires[2]


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Supported by DFG-project HO 3324/4, the German excellence initiative via the "Nanosystems Initiative Munich" (NIM), and the “Center of NanoScience (CeNS)” in Munich.
Efficient electron structure method for type-II broken-gap superlattices

T. Andlauer, T. Zibold, and P. Vogl

Most semiconductors possess positive band gaps in the infrared wavelengths regime. This still holds for most heterostructures that consist of layers of different semiconductors. However, some applications call for a wider tunability of band gaps up to the point where the band gap vanishes. Such a situation can occur if one forms heterostructures of materials with a gross misalignment of their top valence band edge such as shown in Figure 1. Such materials are called type-II heterostructures or materials with “broken” band gaps and well-known examples are InAs:GaSb heterostructures. They are promising candidates for far-infrared devices and possibly for thermoelectric materials.

Up to now, it has been very difficult to quantitatively predict and theoretically analyze such systems. Since broken gap materials strongly mix states of the valence and conduction band states of the constituent materials, a predictive theory of electronic structure of "gapless" mesoscopic structures requires the treatment of electrons and holes on equal footing. While this is automatically the case in all atomistic approaches, they can handle only small nanostructures with few atoms per unit cell. For mesoscopic devices and realistic heterostructures, the most efficient electronic structure method is the so-called envelope function approach (EFA), but it requires the a-priori separation into negatively charged "electrons" and positively charged "holes".

Fig. 1: Type-II or broken gap heterostructure: the bottom conduction band edge of the left material lies energetically lower than the top valence band edge of the right material.

Fig. 2: Left: Schematic band edge states of an InAs:GaSb heterostructure for layer widths larger than 10 nm. Right: Calculated electronic band structure as a function of the lateral wave vector $k_f$ near the Fermi energy. The band gap is indirect.

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We have developed a novel charge self-consistent electronic structure scheme that is able to efficiently and faithfully predict the electronic structure of type-II heterostructures [1,2]. This method is based on a generalization of the standard EFA and does not require an a-priori classification into electrons and holes. In this method, we remain in the electron framework throughout and occupy all included subbands according to Fermi statistics. Subsequently, we subtract a positive background ionic charge that guarantees charge neutrality. With this procedure, we have calculated the electronic structure and optical transition energies of intrinsic InAs/GaSb superlattices as a function of the layer width.

A schematic result is shown in Figure 2, left hand side. For an InAs:GaSb heterostructure, the electronic confinement leads to an effectively positive gap. However, the role of valence and conduction band states is inverted in this case: the highest occupied states are electron-type, the lowest unoccupied states are hole-type. In Figure 2, right hand side, the calculated lateral band structure is depicted for energies near the Fermi energy that further illustrates this result.

**Fig. 3:** Left: Calculated effective band gap of InAs:GaSb heterostructure as a function of the InAs layer width. Right: Comparison of the present theory with several experiments.

We predict the band structure of InAs:GaSb heterostructures to exhibit a band gap that ranges from the mid-infrared to zero (Figure 3 left). For wider InAs wells, we even find semimetallic behavior. As shown in Figure 3 (right), we obtain excellent agreement with experimental data in a regime where the superlattices exhibit a crossover in the energetic order of the lowest electron-like and the highest hole-like subbands.


Supported by the Deutsche Forschungsgemeinschaft SFB 631, SPP 1285 and the Nano-Initiative Munich (NIM).
Isotope effect on electron paramagnetic resonance of boron acceptors in silicon
Andre R. Stegner, Hiroyuki Tezuka, Till Andlauer, Michael. L. W. Thewalt, Kohei M. Itoh, Martin Stutzmann, and Martin S. Brandt

Electron paramagnetic resonance (EPR) and related techniques have contributed extensively to the understanding of substitutional shallow donors in the different allotropes of silicon throughout the last 50 years. On the contrary, EPR had been ineffective for the study of shallow acceptors for a long time. The reason behind this asymmetry lies in the different structure of the respective dopant ground states. While the electronic ground state of substitutional shallow donors in silicon is $s$-like and only twofold spin degenerate, shallow acceptors states have $p$-character and show a fourfold degeneracy. This latter degeneracy can partially be lifted by any symmetry breaking perturbation. Such perturbations, e.g. electric fields or strain, can strongly alter or even dominate the level scheme of acceptor Zeeman energies (inset in Fig. 2) for external magnetic fields that are typically used for EPR measurements. If the perturbation is not homogeneous across the sample, this can easily lead to an extreme inhomogeneous broadening of the EPR resonances. Therefore, it was only in 1978, when Si became available with sufficient crystalline quality, that EPR of B acceptors in externally unstrained Si (Si:B) was reported for the first time. In particular, a correlation between the linewidths of the different boron-related EPR resonances and the concentrations of C and O point defects was established.

However, a number of fundamental questions have remained unsolved: (i) Although a random strain distribution that is induced by point defects should lead to a purely Lorentzian broadening of the EPR resonances, the experimentally obtained lines could only be fitted taking into account a significant Gaussian contribution that was found to be independent of the concentrations of C and O. The origin of this additional broadening, which dominates the overall linewidths in samples with small point defect concentrations and which shows a large angular dependence, has essentially remained unclear. (ii) For the samples of highest crystalline purity, the Lorentzian contribution to the Voigt profiles did not fall below a threshold value of 10–mT which, in the model developed by Neubrand, would correspond to an unreasonably high concentration of point defects. (iii) For two of the six B-related resonances, a distinct substructure was observed (cf. Fig. 1). It was proposed that this structure originates from a dynamic effect, however, the specific mechanism remained unknown.

Fig. 1: EPR spectra of the $B \Delta m=1$ resonances measured for $^{29}$Si and $^{28}$Si (open circles), and simulated spectrum (solid curve) based on the isotope perturbation model.

Reference:

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2Keio University, Yokohama, Japan.
3Simon Fraser University, Burnaby, Canada.
In Fig. 1 we compare EPR spectra of the boron Δm=1 resonances as measured on a high purity sample with a natural isotope composition (natSi) and an isotopically purified ²⁸Si sample. For the latter, we observe more symmetric lineshapes and significantly reduced linewidths, which suggests that the presence of the different Si isotopes is the origin of the unknown broadening effects previously observed for natSi. The observation of the substructure line in ²⁸Si established that (iii) cannot be explained by an isotopic effect.

As detailed in [1], we have set up a theoretical model that allows for a quantitative description of the isotopic broadening effect of acceptor EPR in Si, accounting for the isotopic perturbation via local fluctuations of the valence band edge at ²⁹Si and ³⁰Si lattice sites. The solid curve in Fig. 1 shows a simulation of the Δm=1 spectrum based on our model. As can be seen, we can reproduce the experimentally observed line shapes of the B-related resonances in natSi with excellent agreement between theory and experiment, without the necessity to invoke hyperfine interaction with the B nucleus or with ²⁹Si. A comparison of the experimental and theoretical data in Fig. 2 shows that also the angular dependence of the broadening can be understood quantitatively from our calculations. The open questions raised above can be answered as follows: (i) The Gaussian contribution to the line shape of the broad Δm=1 resonance previously employed to explain the experimentally observed lineshape, is merely a phenomenological way to account for the broadening induced by the isotopic randomness. The 10 mT threshold of the Lorentzian contribution to the line shape (ii) was an artifact that resulted from an inappropriate model of the line shape in the limit of low point defect concentrations.

Further, our calculations show that the perturbation due to the random distribution of the different Si isotopes also leads to a shift in the effective g-values of different B-related resonances, which we could as well verify in our experiments. Concerning issue (iii), our results suggest that the substructure lines of the Δm=1 and Δm=2 resonances originate from a subensemble of acceptors, where the Zeeman level scheme is perturbed such that the |1⟩↔|2⟩, |3⟩↔|4⟩ transition energies are equal. We have discussed these substructure lines and their dynamics in more detail in [2]. Finally, a comparison of our calculation with previous works investigating the B acceptor ground state in the absence of an external magnetic field, provides an independent verification of the energy offsets between the valence bands of the different isotopes of silicon.


Dephasing and decoherence in phosphorus doped silicon

Felix Hoehne\(^1\), Jinming Lu, André R. Stegner, Martin Stutzmann, and Martin S. Brandt

Electron spin resonance (ESR) is a well known tool to manipulate electron spins in semiconductors. Due to the limited detection sensitivity of ESR, typically electrical and optical detection schemes of spin resonance (EDMR, ODMR) are favored to detect small numbers of spins. This becomes important especially in semiconductor nanostructures used e.g. in the context of quantum computation and electronics. Here, we use pulsed electrically detected magnetic resonance (pEDMR) to study dephasing and decoherence properties of \(^{31}\text{P}\) and Si/SiO\(_2\) interface defect spins in phosphorus doped silicon with a \(^{31}\text{P}\) concentration of \(10^{17}\text{cm}^{-3}\). In this system, the spin-to-charge transfer is achieved via a spin-dependent recombination process between the \(^{31}\text{P}\) and the Pb\(_{0}\) spins \([1]\).

We employed the \(\pi/2 – \tau – \pi/2\) Ramsey pulse sequence to characterize dephasing in our samples. \(\pi/2\) denotes a microwave pulse which is adjusted to turn resonant spins by an angle \(\pi/2\) in the Bloch sphere and \(\tau\) denotes a variable evolution time. The results of such a measurement are shown in Fig. 1 a) where the integrated current transient is plotted as a function of the magnetic field and the evolution time \(\tau\). Characteristic oscillations on the time axes and the magnetic field axes are clearly visible around the resonance fields at \(B_0 = 350.7\) mT for the high-field \(^{31}\text{P}\) hyperfine line and, to a lesser extent due to spectral overlap, around \(B_0 = 346\) mT for the low-field \(^{31}\text{P}\) and Pb\(_{0}\) resonances. For comparison, a simulated spectrum assuming a simple two spin model including the inhomogeneous broadening of the \(^{31}\text{P}\) and Pb\(_{0}\) lines is shown in Fig. 1b). From the exponential decay of the oscillations on the time axis we can extract a dephasing time of \(63\pm9\) ns for the high-field \(^{31}\text{P}\) hyperfine line corresponding to a linewidth of \(0.27\) mT which is in good agreement with the value measured in continuous wave EDMR.

We also characterized the transversal relaxation time \((T_2)\) using a \(\pi/2 – \tau – \pi – \tau - \pi/2\) or spin echo pulse sequence and the longitudinal relaxation time \((T_1)\) using a modified inversion recovery pulse sequence. The results are shown in Fig. 2 a) where the amplitudes of the spin echoes are plotted as a function of the evolution time \(\tau_1 + \tau_2\). The data can be described by a monoexponential decay with a time constant of \(\tau_{\text{echo}} = 3.5\pm0.1\) µs. The results of the inversion recovery type of experiment depicted in Fig. 2 b) also show a monoexponential decay with a time constant of \(\tau_{\text{invRec}} = 4.4\pm0.1\) µs. Since \(T_1\) values for \(^{31}\text{P}\) spins at liquid He temperatures are reported to be orders of magnitude longer than the measured value of \(\tau_{\text{invRec}}\), we attribute the observed time constant to the recombination time of singlet \(^{31}\text{P}\)-Pb\(_{0}\) pairs. If we assume that transversal relaxation and singlet recombination are independent processes, the echo decay time will be given by \(1/\tau_{\text{echo}} = 1/\tau_{\text{invRec}}+1/T_2\). Using this relationship, we can extract a transversal relaxation time \(T_2 = 17\pm4\) µs.

This value has to be compared with the \(T_2\) times measured in bulk samples with ESR \([2]\). A linear decrease of the \(T_2\) times from 300µs to 100 µs with increasing \(^{31}\text{P}\) concentration is reported in the concentration range between \(7\times10^{16}\text{cm}^{-3}\) and \(1.2\times10^{17}\text{cm}^{-3}\) which is attributed to spin-spin coupling between the \(^{31}\text{P}\) spins. A much stronger decrease is predicted for higher concentrations with a measured \(T_2\) time of 5µs for a concentration of \(2\times10^{17}\text{cm}^{-3}\). Our value of 17 µs for a sample with nominally \(9\times10^{16}\text{cm}^{-3}\) donors falls within

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this range. However, to confirm this interpretation samples with different $^{31}$P concentrations have to be measured.

![Graph](image)

**Fig. 1:** (a) Integrated current transient after the Ramsey pulse sequence as a function of magnetic field $B$ and waiting time $\tau$. (b) Simulation based on a spin-pair model including inhomogeneous broadening of the lines.

![Graph](image)

**Fig. 2:** Decay of the spin echo (a) and inversion recovery signal (b) as a function of the waiting time. Using the measured values of the decay time constants we can extract a transversal relaxation time of $T_2 = 17\pm 4$ µs.


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Supported by Deutsche Forschungsgemeinschaft, Collaborative Research Center 631, C3
Towards efficient silicon-based light sources using tailored photonic materials

Norman Hauke\textsuperscript{1}, Thomas Zabel, Stefan Lichtmannecker, Michael Kaniber, Dominique Bougeard, Jonathan Finley, Gerhard Abstreiter, and Yasuhiko Arakawa

The development of silicon based photonics lags far behind III-V semiconductor materials, despite silicon nano-fabrication methods being extremely well developed. The reason is the indirect-band-gap of silicon for which light emission is a phonon mediated process with a low probability. This gives rise to spontaneous e-h recombination lifetimes up to the millisecond range, much longer than the typical timescale for non-radiative recombination as Auger-recombination. As a result, the internal quantum efficiency for intrinsic silicon is very low with typical values being around \(10^{-6}\). In our devices, the internal quantum efficiency is enhanced by combining quantum size effects with cavity-QED phenomena in photonic crystal nanocavities. This nano-photonic approach has already been shown to be highly effective for III-V semiconductors.

We report the fabrication and optical investigation of nanophotonic light sources based on passive Silicon-on-Insulator photonic crystal (SOI-PC) nanostructures \cite{1}. The samples investigated were fabricated from SOI wafers with a 250nm thick active silicon layer on top of a 3\(\mu\)m thick layer of SiO\textsubscript{2}. A 2D PC consisting of a triangular lattice of air holes with a period of \(a=275\text{nm}\) is patterned into the upper silicon layer. This was done using electron-beam lithography and subsequent reactive-ion etching (RIE). As a final fabrication step, the underlying SiO\textsubscript{2} is removed by hydrofluoric acid (Fig 1(a)). Low mode-volume nanocavities were realized by omitting three air holes in a row and by shifting the lateral holes away from the cavity center by 0.15 lattice constants to form modified \(L_3\) cavities (Fig 1(b)). Our investigations clearly show that the photonic crystal fabrication process results in the appearance of a broad defect emission band ener-

\begin{figure}[h]
\centering
\includegraphics[width=\textwidth]{Figure1.png}
\caption{(a) Left panel: schematic cross-sectional representation of the PC nanocavity structures investigated. Right panel: layer sequence in the active region. (b) Left panel: SEM image showing a PC from top. Right panel: zoom-in to the modified \(L_3\) defect where the outer holes are shifted by 0.15 lattice constants to the outside.}
\end{figure}

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The intensity of which is significantly enhanced by coupling to the photonic defect nanocavity mode, labeled M1. Spatially resolved spectroscopy demonstrates that this defect band is present only in the region where air-holes have been etched during the fabrication process. Pronounced photoluminescence emission is obtained from the SOI-PC defect cavity. The intensity of the emitted light is >10x stronger from the photonic defect nanocavity as compared to unpatterned regions of the device. Analysis of our results indicates that mainly the phonon satellites of the interband silicon emission and also, but more weakly, surface defect states are responsible for the luminescence of PC cavity modes.

![Figure 2](image-url)

**Figure 2:** (a) PL spectra as a function of temperature for a PhC with r/a = 0.34. (b) Arrheniustype analysis for the TO replica (squares), defect band emission (circles) and M1 mode emission (triangles). The activation energies extracted from linear fits label the three curves.

Temperature dependent microphotoluminescence investigations of SOI-PCs shows that detectable emission from the cavity mode persists up to room-temperature (Fig 2(a)). In strong contrast, the background emission vanishes for temperatures higher than 150 K. An Arrhenius type analysis (Fig 2(b)) of the temperature dependence of the luminescence recorded either in resonance with the cavity mode, or spectrally detuned, suggest that the higher temperature stability arises from enhanced internal quantum efficiency due to the Purcell-effect [1].

In conclusion, we presented a temperature-dependent investigation of PL in Silicon PhC-nanocavities. We suggest two mechanisms that are responsible for the luminescence of cavity modes: mainly phonon-mediated recombination from charge carriers in the electronic band states and also, but more weakly, recombination from charge carriers trapped in surface defect states. Furthermore, we observed an enhanced internal quantum efficiency in spectral resonance with the cavity mode emission.


Supported by the TUM International Graduate School of Science and Engineering (IGSSE) and the TUM Institute for Advanced Studies (IAS)
Manipulation of light using nanoscale plasmonic structures

Gregor Bracher, Birgit Wiedemann, Simon Frédérick, Michael Kaniber, and Jonathan Finley

The need to bridge the gap between the electronics and photonics in future information technologies, have sparked much interest in the manipulation and control of light on the nanoscale using surface plasmon polaritons. Surface plasmon polaritons (SPPs) are collective oscillations of the conduction band electrons at a metal-dielectric interface and offer a strong possibility to localize, guide and manipulate light in strongly sub-wavelength sized metallic structures.

Over the past year we have actively investigated two different aspects of SPPs; namely (i) the enhancement of the luminescence intensity of colloidal nanocrystals brought into the vicinity of metallic nanoparticle arrays and (ii) the opportunity to use SPPs to guide light on a chip in sub-wavelength metallic nanowires. In this report, we describe recent activities in both directions.

(i) For such studies nanometer sized metallic objects such as triangles or spheres are arranged in periodic lattice (see inset of Figure 1). Depending on the size, shape, lattice symmetry and separation of the structures, different dipolar and quadrupolar plasmonic resonances can be excited and manipulated. In the inset of Figure 1 we show a scanning electron microscope image of a nano-triangle array with triangle base length and lattice constant . In Figure 1 we present the typical results of white light transmission measurements of nano-triangles with a different lattice constant but constant nano-triangle size. A clear shift of the peak of the scattered light to longer wavelengths is observed as the lattice pitch increases, due to a change of the quadrupolar resonance of the nano-triangle array.

In other studies (ii), we probe the propagation of surface plasmon polaritons along thin metallic nanowires defined using electron beam lithography and Au evaporation on GaAs substrates. An example of such structures is depicted schematically in figure 2(a), together with an image of structures fabricated in our laboratories in the WSI. The typical lateral size of such nanoscale plasmonic waveguides is ~100nm, the Au film has a thickness of ~100nm and the waveguide length is up to 100µm.

Figure 1: Transmission measurement revealing the plasmonic resonance. With increasing pitch of the triangles, the resonance is shifted towards longer wavelengths. Inset: SEM image of a square pattern of nano-triangles.

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There are several possibilities to overcome the wavevector mismatch and launch plasmons from free space into such structures. Examples include grating coupling [1], prism coupling [2] or end-fire coupling[3]. First experiments in these directions have relied on the end-fire excitation method, since one can readily control the polarization and the intensity of the excitation incident on the end facet. Control of the polarization is one of the key elements since only p-polarized light, with an electric field perpendicular to the metal-dielectric interface, is capable of generating propagating SPPs. Once the SPP is launched, it propagates along the wire and can be detected, when it scatters at the end of the wire, as shown by the camera image shown in figure 2 (b). In figure 2(c) we show the intensity of the light detected at the end of the plasmonic waveguide as a function of the polarization of the excitation source. The degree of polarization is typically >98% p-polarization. This value is much higher than the degree of polarization of a dielectric scatterer (15%) or a simple metallic cylindrical scatterer (61%) on the sample surface. These observations confirm that we do indeed launch, guide and detect SPPs from the far field in such nano structures. Recent measurements indicate that, for SPPs excited at $\lambda=900$nm on a nanowire with a cross-section of 100x100nm$^2$, the propagation length is >5µm. These observations, coupled with our newly established capabilities to perform 2-axis confocal microscopy measurements at low temperatures provide significant promise for future experiments where we will attempt to couple spatially separated, individual quantum emitters using such metallic nanostructures.


We acknowledge financial support of the DFG via the SFB 631, Teilprojekt B3, and the German Excellence Initiative via NIM.
Non-resonant feeding of photonic crystal nanocavity modes by quantum dot multi-exciton states

Michael Kaniber¹, Arne Laucht, Abbas Mohtashami, Thomas Günthner, Max Bichler, and Jonathan J. Finley

Individual semiconductor quantum dots are known for more than ten years as promising candidates for deterministic generation of single photons on demand. Combining those “artificial atoms” with nanoscale solid-state optical cavities is highly desirable in order to control its spontaneous emission properties as well as to strongly enhance its photon extraction efficiency. The latter might finally clear the way for single photon turnstile devices being useful in every day data communication.

Photoluminescence studies of photonic crystal defect nanocavities containing single quantum dots typically reveal intense emission from the cavity mode, even when the discrete quantum dot transitions and the photonic crystal cavity mode are strongly spectrally detuned (|ΔE|>20 meV) [1,2,3]. This has been unambiguously shown by applying photon cross-correlation spectroscopy between the quantum dot and the far detuned cavity mode, demonstrating that the cavity mode is fed by the same quantum dot.

For small dot-cavity detunings (|ΔE|<5 meV), acoustic phonon mediated dot - cavity coupling gives rise to emission into the cavity mode [6,7]. However, for larger detunings up to 20 meV acoustic phonon mediated coupling becomes ineffective and additional effects come into play. Over the past year we demonstrated experimentally that such off-resonance emission from the cavity mode is related to multi exciton-continuum transitions during the cascaded emission from the dot [4] and, thus, finally understood a heavily discussed mystery in the field of solid-state cavity quantum electrodynamics.

Therefore, we applied spectrally resolved time-correlated single photon counting spectroscopy to track the temporal evolution of the emission spectrum. We find that the mode emission is temporally correlated with the quantum dot multi-exciton emission but becomes much weaker after the population in the dot reduces towards the single exciton level [4]. As shown in Figure 1, we observe that the cavity mode is indeed temporally correlated with the quantum dot p-shell emission. In strong contrast, the quantum dot s-shell emission is temporally delayed due to the cascaded emission.
sion characteristics. This general result observed on many different dot-cavity systems for different spectral detunings lends strong support to recent pump power dependent cross-correlation measurements [5] identifying such multi exciton-continuum transitions as feeding source of the cavity mode.

In addition, pump power dependent time-resolved measurements performed on the dot-cavity system lend further support to our identification of non-resonant cavity feeding and allow us to extract quantitative (we do not have any quantitative information about the coupling strength…) information about the strength of the off-resonant coupling.

Since the cavity mode is fed via optical transitions between multi-exciton states and a quasi-continuum of multi-exciton states [4], this naturally results in several photons at the cavity mode frequency per excitation cycle. It manifests itself as a strongly enhanced number of correlations when probing the photon statistics of high quality factor, low mode volume solid-state nanocavities using a Hanbury Brown and Twiss intensity correlation setup.

This strong increase of the multiphoton emission probability at zero delay time reduces the purity of the single photon source and, thus, lowering its usefulness for generating a highly efficient and deterministic stream of individual photons with a high degree of quantum indistinguishability. The latter is a key requirement for many novel applications in quantum information processing, including quantum key distribution and the realization of quantum computing based solely on linear optics components.

Moreover, obtaining an understanding of this non-resonant interaction in semiconductor cavity systems is essential for the optimization of low threshold nano lasers and in the use of cavity QED phenomena in solid-state quantum optics.


We acknowledge financial support of the Deutsche Forschungsgemeinschaft via the Sonderforschungsbereich 631, Teilprojekt B3 and the German Excellence Initiative via “Nanostems Initiative Munich”
Spin effects in single self-assembled InGaAs quantum dots

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The spin of a single electron confined in a quantum dot (QD) is a promising candidate for future quantum information applications and has attracted much attention over the last 20 years [1]. Over the past year we have managed to demonstrate all optical preparation and readout of a single electron spin in an individual self-assembled InGaAs quantum dot (QD). Our methods combine spin to charge conversion with luminescence recycling [2] and are applied to optically probe the spin relaxation dynamics. Importantly, our approach allows us to probe the dynamics of a single spin over ultra long timescales (≥200 μs), generally inaccessible to optical single spin readout.

The devices investigated are voltage tunable QD spin memory structures that can be switched between two modes of operation; (i) charging, where optically generated holes are removed from the dot whilst electrons remain stored due to the presence of an AlGaAs barrier and (ii) readout, where excitons optically pumped into the dot recombine to produce luminescence [2]. A single electron is prepared in the dot by setting the device to the charging mode and applying a resonant laser pulse tuned to the X\(^0\) transition (Fig. 1-A). The spin of the electron is controlled via the polarization of the preparation laser with high fidelity since the pure spin X\(^0\) eigenstates are not mixed by the anisotropic electron-hole exchange interaction before hole tunneling escape occurs. After generation, the electron is stored and its spin tested after a controlled delay time via the conditional absorption of a second laser pulse tuned to the X\(^-1\) transition. This readout pulse converts the spin of the stored electron into a charge occupancy (1\(e\) or 2\(e\)), which is then repeatedly sampled during the readout phase of the measurement. Our experiments confirm that the dot can be selectively prepared in a 2\(e\) state by successive excitation of the X\(^0\) and X\(^-1\) transitions. The charging and discharging dynamics are probed using time dependent measurements to probe the tunneling escape time of the electron and hole. Furthermore, measurements in high magnetic fields exhibit a clear Pauli blockade in the 1\(e\) → 2\(e\) charging step (Fig. 1-B). This blockade can be lifted by delaying the 2\(e\) charging pulse by times

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longer than the electron spin $T_1$ time (Fig. 1-C,D). The spin lifetime of the stored electron was measured by monitoring the storage time dependence of the blockade as the temperature was varied. We observe a $T_1 \sim T^{-0.87\pm0.16}$ dependence, very close to the expected $T^{-1}$ for spin relaxation due to phonon mediated spin-orbit interaction. From the equilibrium degree of polarization of $X^{-1}$ and its dependence on temperature an absolute value of $g_e = 0.86\pm0.03$ was extracted, in excellent agreement with $\mathbf{k} \cdot \mathbf{p}$ calculations for our QDs.

Almost all advances have been made in III-V materials in which the decoherence time that gives the time for logical quantum operations is predominantly limited by the hyperfine coupling of the electron spin with the spins of the $10^4$-$10^5$ resident nuclei in the QD. Therefore, it is of great interest to gain a better understanding of this complex effect. We also investigated novel unipolar optically pumped dynamic nuclear polarization (DNP) dynamics in a single self-assembled InGaAs quantum dot (QD). Photocurrent measurements are performed, whereby spin oriented excitons are resonantly pumped into the lowest orbital state of the QD. They remain in the dot for a time ($<1\text{ns}$) during which the electron interacts with the nuclear spin system via the hyperfine contact coupling. This leads to electron-nuclear flip-flop processes that can polarize the nuclear spin bath, creating an Overhauser magnetic field $B_N$ that causes an energy shift of the exciton states due to the additionally introduced Zeeman interaction of the electron with $B_N$ (see Fig. 2). Remarkably, we observe a strong asymmetry in nuclear spin pumping for excitation of the two Zeeman-split neutral exciton states. Hereby, pumping the higher energy Zeeman branch effectively polarizes the nuclear spin system, whereas the lower energy branch does not (see fig.1). This surprising behavior is found to persist over the whole parameter range for which DNP occurs. We also find a characteristic dependence of the observed DNP on $B_{\text{ext}}$. For very low ($<1\text{T}$) and very high ($>8\text{T}$) magnetic fields, the steady state Overhauser field is strongly suppressed. Optically induced nuclear spin pumping is most efficient for an intermediate magnetic field regime of $B_{\text{ext}} 4 – 6 \text{T}$, with a magnitude of $B_N$ up to 3.7 T corresponding to a 53% polarization of the nuclear spin bath. This behavior can be understood employing a model based on the bright and dark exciton level structure of the system, consisting of two bright exciton states that are resonantly excited in our experiment, two optically inactive dark exciton states, and the crystal ground state of the QD. Electron-nuclear spin flip-flop processes transfer the exciton from a bright into the corresponding dark state at a rate that strongly depends on the energy gap involved in the transition which varies as a function of $B_{\text{ext}}$. The magnetic field dependence of DNP and the observed asymmetry was found to appear naturally in our model when the master equation of the system is solved for typical QD parameters.

Theory of electron energy relaxation and decoherence in self-assembled quantum dots

Thomas Grange

Understanding energy relaxation and decoherence in semiconductor quantum dots is of prime importance for the design of new optoelectronic devices and for their potential application in quantum information.

Mechanisms of electron relaxation and decoherence in self-assembled quantum dots have been theoretically investigated. The problem of energy relaxation in quasi zero-dimensional systems has been a subject to debate since the early 90’s and the prediction of the so-called phonon bottleneck effect. Indeed, owing to the discrete energy levels in these nanostructures, emission or absorption of optical phonon is impossible unless the energy separation between two electronic levels match the one of optical phonons.

Nevertheless, this phonon bottleneck has never been clearly observed. In fact, electrons and optical phonons are in the strong coupling regime in these structures, so that taking into account the damping of optical phonons due to the crystal anharmonicity allows to explain the energy relaxation between two discrete energy levels.

We have developed a new theoretical approach to describe the electron dynamic in quantum dots, taking into account the strong coupling to optical phonons as well as a microscopic description of anharmonic instability of these optical phonons [1]. Recently, an experimental study of intraband transition dynamics using terahertz pump-probe measurements [2] has demonstrated the ability of this model to explain and predict electron dynamic in quantum dots. In particular, a strong increase of the lifetime has been evidenced when the transition energy is decreased below the LO-phonon energy, with lifetimes up to the

Fig. 1: Relaxation time between the first excited p state and the ground state as a function of the transition energy. Calculations in solid line are compared to measurements in symbols [1,2].

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nanosecond (fig. 1), orders of magnitude longer than in quantum wells. This result highlights the great potential of quantum dots for terahertz applications.

Except at very low temperature, the homogeneous linewidth of such intraband transition is not limited by its sole population relaxation time, but also by pure dephasing effects, as evidenced by photon echo experiments [3]. In general, acoustic phonons which are efficiently coupled to electronic states have too little energy to be resonantly absorbed or emitted by quantum dot electronic transitions. Nevertheless, they can trigger virtual transitions which do not need to fulfill energy conservation, provided that the virtual state lifetime is on the very short timescale allowed by the energy-time uncertainty principle.

We have recently developed a nonperturbative calculation of the dephasing due to both real and virtual acoustic phonon-assisted transitions with higher excited states [4]. This approach gives nice agreements with experiments [3] and predicts the intrinsic limitations of coherence properties of quantum dot optical transitions.

**Fig. 2:** Calculated intraband absorption from the ground state to the first excited state of a QD for different temperature.


Supported by the Alexander von Humboldt-Foundation
Electrostatically defined few electron double quantum dot based on Si/SiGe heterostructures

Andreas Wild¹, Jürgen Sailer, Joachim Nützel, Gerhard Abstreiter, Dominique Bougeard

Coupled electrostatically defined quantum dots (QD) are presently intensely investigated as possible spin qubits for quantum information processing (QIP). The electron spin coherence time in the currently most advanced systems, which are GaAs based, is limited by the hyperfine interaction of a confined electron with the many nuclear spins of the surrounding host crystal. Using Silicon (Si) as the host material offers a promising alternative route towards solid state based QIP. For Si in its natural isotopic composition, substantially longer decoherence times are expected compared to GaAs as a result of reduced hyperfine interaction and weak spin-orbit coupling. The latter originates from a lower atomic mass of Si and no piezo-electric interaction. The possibility of isotopic purification of the Si crystal to contain no nuclear spins promises even superior coherence properties for electron spins. Finally, the mature Si technology opens up the opportunity of rapid industrial scalability.

We choose strained Si in SiGe as the host material for laterally confining electrons within a two-dimensional electron gas (2DEG). This approach enables high electron mobilities and allows for isotopic engineering of the host crystal. By using single crystals enriched in certain isotopes as source materials in our molecular beam epitaxy (MBE) system, we are able to create virtually nuclear spin free hosts or to decorate the host crystal lattice with a variable concentration of nuclear spins to investigate the influence on qubit decoherence for prospective quantum dot circuits.

We have established the Si/SiGe heterostructure development and processing of 2DEGs and devices suitable for the definition of coupled QDs.

A typical MBE grown layer structure is depicted in Fig. 1a). The 2DEG forms in the Si quantum well (QW) which is separated from the modulation doped phosphorous (P) region by the SiGe spacer layer. We have realized both, 2DEGs in natural Si and isotopically purified ²⁸Si crystals [1] whose sheet carrier densities can be tuned via front and back gates from 0 to 5·10¹¹ cm⁻². We achieve 2DEG mobilities up to 110,000 cm²/(Vs) for Si/SiGe in its natural isotopic composition and up to 60,000 cm²/(Vs) for ²⁸Si/SiGe 2DEGs. On top of the SiGe and Si capping layers, nanostructured palladium (Pd) top gates can be evaporated that are used to deplete the 2DEG underneath.

Based on these 2DEGs we have then defined and characterized first double QDs in natural Si as shown in Fig. 2b). The AFM micrograph presents a typical gate design of our double QD devices. An artificial QD molecule forms in the center and can be controlled by applying different gate voltages to the confining plunger gates. It is furthermore connected to the electron reservoirs III and IV.

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**Fig. 1:** (a) Schematic heterostructure layout. (b) AFM micrograph of a typical double QD gate design.
via tunable tunneling barriers between gate bL or bR and C. By measuring the conductivity between contacts IV and V, we can use another QD between gate bR and xR as a charge sensor for the nearby double QD. The conductivity through this sensor QD is extremely sensitive to the local electrostatic environment and thus to the occupation of the double QD.

The sensor current depicted in Fig. 2a) can be directly mapped to different occupations of the double QD. (b) and (c) Effect of a rectangular pulse sequence applied to gate bL (b) and bR (c) on the QD occupation. (Inset) Energy level structure of a double QD as the energy level of the left QD is pulsed between two configurations above and below the chemical potential of the lead.

![Fig. 2: (a) Charge stability diagram in the vicinity of an anti-crossing of two charging lines. The charge sensor current can be directly mapped to different occupations of the double QD. (b) and (c) Effect of a rectangular pulse sequence applied to gate bL (b) and bR (c) on the QD occupation. (Inset) Energy level structure of a double QD as the energy level of the left QD is pulsed between two configurations above and below the chemical potential of the lead.](image)

In summary, this study demonstrates the suitability of Si/SiGe double QDs for future coherent spin experiments and as spin qubits.


This work has been performed in cooperation with Stefan Ludwig, Fakultät für Physik and Center for NanoScience, LMU. We acknowledge funding by the DFG via SFB631.
Dynamics of indirect excitons in low-dimensional potential landscapes

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Photogenerated electron-hole pairs in double quantum well devices can be manipulated both in lifetime and position via a mesoscopic voltage-controlled electrostatic landscape [1]. The quantum-confined Stark effect allows us to create long-living indirect excitons with a microsecond-lifetime. Recently, we demonstrated a novel electrostatic trap for indirect excitons in coupled GaAs quantum wells embedded in a field-effect device [2]. There, the indirect excitons are trapped in the quantum wells just below the perimeter of SiO\textsubscript{2}-layers, which are sandwiched between the surface of the GaAs heterostructure and a semi-transparent metallic top gate.

In a recent study, we investigated the lateral expansion of dipolar excitons in coupled quantum wells in two (2D) and one (1D) dimensions (see Figure and ref.[3]). In 2D, the exciton expansion obeys nonlinear temporal dynamics due to the repulsive dipole pressure at a high exciton density. In contrast, the observed 1D expansion behaves linearly in time even at high exciton densities. The corresponding 1D diffusion coefficient exceeds the one in 2D by far and depends linearly on the exciton density. The findings are attributed to screening of quantum well disorder by the dipolar excitons in one dimension.

\begin{figure}[h]
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\includegraphics[width=\textwidth]{figure1.png}
\caption{Spatial distribution of the photoluminescence of indirect excitons in a two-dimensional (2D) and one-dimensional potential landscape in a GaAs-based heterostructure at 6 K [3].}
\end{figure}


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Supported by DFG Project No. KO-416/17-2, the Center for NanoScience (CeNS), and the German excellence initiative via “Nanosystems Initiative Munich (NIM)” and “LMUexcellent.”
Material Science
P-doping mechanisms of MBE grown GaAs nanowires

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Semiconductor nanotubes and nanowires have stimulated a new enthusiasm in the emerging field of nanotechnology by enabling extraordinary progress in applications and in fundamental mesoscopic physics. GaAs nanowires are a particularly interesting material system because of the possibility of extending the technology of mesoscopic physics directly in a one dimensional system. Molecular beam epitaxy (MBE) can particularly contribute to this, thanks to the high purity and enablement of high quality axial and coaxial heterostructure fabrication. Doping of nanowires is an essential element for the realization of electronic and optoelectronic devices. The doping mechanisms of nanowires have been deeply investigated in the case of Vapor-Liquid-Solid growth of silicon nanowires by Chemical Vapor Deposition (CVD). Indeed, it has been shown that differences in the precursor decomposition rates between the solid nanowire surface and the catalyst lead to the formation of a doped shell and a much lower doped core. Doping of III-V semiconductor nanowires has been demonstrated in the case of inclusion of magnetic impurities and by the inclusion of zinc in the growth process. Electrical doping and transport properties of GaAs nanowires have not been extensively investigated yet, mainly for two reasons: (i) the challenges in doping and (ii) the existence of a Schottky barrier between GaAs and the metal, which make the realization of ohmic contacts challenging specially at low temperatures.

In this work, we have investigated the doping mechanisms of catalyst-free MBE grown GaAs nanowires. For this, have simultaneously measured the spatial dependence of the dopant concentration and the spatial dependent resistivity, respectively by Raman spectroscopy and four point electrical measurements. A schematic drawing of this experiment is shown in Fig.1. Several samples obtained under different growth conditions have been investigated. This study has allowed us to reveal that two competing mechanisms exist, as shown in Fig.2: dopant incorporation from the side facets and from the gallium droplet. Moreover, for the incorporation through the droplet, we have seen that doping compensation seems to play an important role in the carrier concentration. By varying the incident silicon (dopant) flux and the nanowire growth rate, different carrier concentrations have been obtained. Hole concentrations of at least 2.4x10¹⁸ cm⁻³ have been achieved [1,2], which to our knowledge is the largest p doping range obtained up to date.

Fig. 1: Schematic drawing of the experiment combining Raman spectroscopy and electrical characterization

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The possibility of extremely high doping in GaAs nanowires opens the avenue for using these nanowires in low temperature mesoscopic physics experiments and modern nanoscale devices. More detailed studies that also combine a complete structural analysis are currently under way, for a better estimation of the carrier concentration and of fundamental properties such as carrier mobilities and diffusion lengths.

**Fig. 2:** Schematics of two extreme doping mechanisms in nanowires: a) through side facet deposition and b) through the droplet.


Raman spectroscopy of wurtzite and zinc-blende GaAs nanowires

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Semiconductor nanowires are expected to play an important role in several areas of technology, such as electronics, sensing and energy conversion. The functional properties of nanowires are influenced by different factors, the structure being one of the most determining. Raman spectroscopy is a non-destructive technique used for the structural characterization of materials, which enables to determine the crystalline phase and presence of strain.

In this work, the structural properties of gallium arsenide (GaAs) nanowires have been studied by spatially resolved Raman spectroscopy, and the obtained results are compared with Transmission Electron Microscopy measurements. The GaAs nanowires present a mixture of zinc-blende (ZB) and wurtzite (WZ) structure [1,2]. Polarization dependent Raman scattering experiments realized on single GaAs nanowires with different percentages of zinc-blende and wurtzite structure allowed the determination and of the selection rules for the special case of nanowires [3].

In the case of ZB, the transversal optical mode \( E_{1}(TO) \) at 267 cm\(^{-1} \) exhibits the highest intensity when the incident and analyzed polarization are parallel to the nanowire axis, as found by the Raman measurements presented in Fig. 1. This is a consequence of the nanowire geometry and dielectric mismatch with the environment, and in quite good agreement with the Raman selection rules. We also find a consistent splitting of 1 cm\(^{-1} \) of the \( E_{1}(TO) \).

![Fig. 1: (a) Raman scan from a nanowire crystallized in ZB structure, obtained using exciting light polarized along the nanowire axis; (b) A series of parallel and perpendicularly polarized Raman spectra obtained using exciting light polarized parallel and perpendicularly to the nanowire axis, from a nanowire with 100% ZB structure. The spectra have been shifted vertically; (c) azimuthal dependence of the \( E_{1}(TO) \) mode related to the ZB structure in the nanowire. Spheres and open squares represent the parallel and perpendicular components of the Raman signal collected with respect to the nanowire axis, respectively. The continuous line is a squared sine fit to the data.](image)

The results of polarization dependent Raman scattering experiments realized on single GaAs nanowires with 30% of WZ phase are presented in Fig. 2. The transversal optical mode related to the WZ structure, \( E_{2}^{H} \), is measured between 254 and 256 cm\(^{-1} \), depending on the wurtzite content. The azimuthal dependence of \( E_{2}^{H} \) indicates that the mode is excited with the highest efficiency when the incident and analyzed polarization are perpendicu-
lar to the nanowire axis, in agreement with the selection rules. Furthermore, the presence of strain between wurtzite and zinc-blende was analyzed by the relative shift of the $E_{1}$(TO) and $E_{2}^{\text{H}}$ modes. For zinc-blende rich nanowires the wurtzite phases exhibit compressive strain (blue shift of the Raman mode) while for wurtzite rich nanowires the zinc blende phases are under tensile strain (red shift of the Raman modes). The existence of strain in wurtzite/zinc-blende structures is in agreement with the slight difference in lattice constants of the two structures.

**Fig. 2**: Parallel polarized Raman scan from a nanowire presenting 30% WZ structure, obtained using (a) perpendicularly polarized incident light and (b) parallel polarized incident light. (c) Representative Raman spectra realized under the main four configurations, from the upper WZ rich segment of a nanowire with 30% of wurtzite structure. The spectra have been normalized and shifted vertically. (d) Representative Raman spectra realized under the main four configurations from the lower ZB rich segment of a nanowire with 30% of wurtzite structure. The spectra have been shifted vertically but the relative intensities can be compared directly; (e) Azimuthal dependence of the $E_{1}$(TO) mode. (f) Azimuthal dependence of the $E_{2}^{\text{H}}$ mode, related to the wurtzite structure. In both polar plots, spheres and open squares represent the parallel and perpendicular components of the Raman signal collected, respectively and the continuous line is a squared sine fit to the data.

Finally, we found an increased intensity of the $A_{1}^{\text{H}}$(LO) phonon mode in correlation with the existence of surface roughness, i.e. $\{111\}$ faceting at the end of the nanowire.


Supported by the Marie Curie Excellence grant ‘SENFED’, SFB 631 and the excellence initiative of the DFG Nanosystems Initiative Munich
Photocurrent and photoconductance properties of a GaAs nanowire

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Semiconductor nanowires have attracted considerable attention for the past few years because of their compelling electronic, mechanical and optical properties. We investigated the optoelectronic properties of photodetectors based on single p-doped GaAs nanowires. The nanowires are electrically contacted to gold electrodes using a versatile focused ion beam (FIB) deposition technique [1]. We experimentally identified two dominating optoelectronic processes in the nanowire photodetectors by measuring the optoelectronic response of the devices as a function of voltage, modulation frequency, photon energy, spatial coordinate, and the polarization of the laser field. On the one hand, a photocurrent is generated at the Schottky contacts between the GaAs nanowires and the metal source-drain electrodes. On the other hand, we observed a photoconductance effect, when illuminating the GaAs nanowire far away from the contacts. We interpret this photoconductance effect to arise from band bending effects caused by surface states on the nanowire surface. In particular, optically generated excess electrons are trapped at the surface, where they act as a negative gating voltage on the p-doped nanowires (photogating effect) [2]. At the same time, the optically excited free excess holes raise the Fermi-energy of the hole gas within the nanowires (photodoping effect) [3]. We demonstrated that the corresponding optoelectronic responses of the nanowires are sensitive to the orientation of linear polarized light. Hereby, the metal-GaAs nanowire-metal circuits act as polarization-sensitive photodetectors, which can be integrated into electronic circuits by the FIB-deposition technique in a very versatile way.

Fig. 1: (a) Scanning electron micrograph (SEM) image of the device. (b) Photocurrent contour plots as a function of the spatial coordinates.

The circuit in Fig. 1(a) features three positions I, II, and III. At position I and III, Schottky contacts between the nanowire and the gold electrodes can be assumed. At position III, however, the nanowire is covered by an opaque carbon layer with a thickness of ~350 nm. At position II, the nanowire is freely suspended. Hereby, the circuit allows locally address-
ing the optoelectronic processes in the middle of the nanowire independently from the ones occurring at the metal-semiconductor contact. We acquired photocurrent images of the nanowire circuit by recording the photocurrent response $I_{\text{PHOTO}}$ at a finite source-drain bias $V_{SD}$, while scanning the laser spot laterally across the sample. As can be seen in Fig. 1 (b) a dominating contribution of $|I_{\text{PHOTO}}|$ occurs at position I compared to $|I_{\text{PHOTO}}|$ at position II.

![Graph](image-url)

**Fig. 1:** (c) Photocurrent response $|I_{\text{PHOTO}}|$ as a function of the angle $\phi$ between the linearly polarized laser field and the orientation of the nanowire for positions I (circles) and II (squares) (b) $|I_{\text{PHOTO}}|$ as a function of $f_{\text{CHOP}}$ for positions I (circles) and II (squares).

The observed photoconductance and -current signals depend on the polarization of the exciting laser field for both positions I and II [Fig. 2(a)]. The signals follow a $\cos \phi$ dependence, where $\phi$ is the angle between the wire and the light polarization. We observe that the polarization ratio larger at position II ($\sim 35 \%$) than close to the Schottky contact of position I ($\sim 15 \%$). In order to test the response time of the metal-nanowire-metal photodetector, we measure $I_{\text{PHOTO}}$ of the devices as a function of $f_{\text{CHOP}}$ [Fig. 2(b)]. For both positions I and II, we do not detect any dependence of $I_{\text{PHOTO}}$ on $f_{\text{CHOP}}$ for frequencies between 0.1 and 6 kHz. This reveals that the effects causing the photocurrent and the photoconductance occur on a timescale shorter than $\sim (6 \text{ kHz})^{-1} = 167 \mu\text{s}.$


Supported by the DFG Grant No. Ho 3324/4, the German Excellence Initiative via the “Nanosystems Initiative Munich (NIM),” Marie Curie Excellence Grant SENFED, and the Swiss National Science Foundation (Grant No. 200021-121758/1) as well as the “Center for NanoScience” (CeNS) in Munich.
Structural and optical properties of high quality zinc-blende/wurtzite GaAs nanowires heterostructures

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Heterostructures consist of the combination of two materials, with different band gaps and electron affinities. In heterostructures of type I, electrons and holes will tend to localize in the lower bandgap material, while in type II these two types of carriers will be spatially separated. Heterostructures are typically formed with chemically different semiconductors, such as GaAs and Al$_{1-x}$Ga$_x$As. In this work, we present a structural analysis as well as the optical properties of GaAs based ‘heterostructures’, whose junction is formed by the same material in the two crystalline phases: zinc-blende and wurtzite [1].

In the bulk state, GaAs is stable in the zinc-blende structure. When reduced to a nanoscale volume, such as in the form of nanowire, wurtzite structure becomes stable. An example of such a zinc-blende/wurtzite heterostructure is given in Fig. 1. High resolution transmission electron microscopy (HRTEM) micrograph from a part of a nanowire is presented in Fig. 1 (a). Here, we used a color code for denoting the different crystal phases: blue for zinc-blende and red for wurtzite. A high angle annular dark field (HAADF) image is superimposed on the micrograph. For clarity, this is plotted enlarged in Figure 1 (b) where the atomic arrangement is also schematized. Fig. 1 (c) presents schematically the theoretical band alignment in a zinc-blende/wurtzite heterostructure according to [1].

**Fig.1:** (a) HRTEM micrograph from a part of a GaAs nanowire exhibiting both zinc-blende (blue circles) and wurtzite structure (red circles). (b) HAADF image from the boxed region in (a). The atomic arrangement is also schematized. (c) Theoretical band alignment at the zinc-blende/wurtzite interface according to [1].

The optical properties of the zinc-blende and wurtzite heterostructures were investigated by photoluminescence (PL) and cathodoluminescence (CL) spectroscopy at 4.2 K and 10 K respectively. A typical PL spectrum from a single nanowire exhibiting a variety of zinc-blende and wurtzite quantum wells with different thicknesses is presented in Fig. 2 (a). Here, besides the emission corresponding to the free exciton recombination in zinc-blende GaAs (E=1.515 eV) several sharp emission peaks at energies between 1.460 eV and 1.515 eV appear. We have investigated the PL emission from several nanowires exhibiting polytypic zinc-blende/wurtzite structure. The PL spectra are characterized with several emission energies in the range of 1.436 eV-1.515 eV depending from the total wurtzite content. [2]. However, the exact position of the peaks differs from nanowire to nanowire and along a single nanowire as well, due to the different sequences of zinc-blende and
Supported by the Marie Curie Excellence grant ‘SENFED’, SFB 631 and the excellence initiative of the DFG Nanosystems Initiative Munich

In order to illustrate this we performed spatially resolved PL measurements on single nanowires. Fig. 2 (b) shows the spatially resolved PL spectrum from the same nanowire presented in Fig. 2 (a). Here, it is clearly observed that the free exciton emission is present only in one part of the nanowire, in accordance with TEM analysis, while the other emission energies are localized along the nanowire length [2]. Improved spatial resolution can be obtained with CL spectroscopy. In Fig. 2 (c) a composite CL image containing three different emission energies is given. As can be seen, the different emissions originate from different locations in the nanowire. These characteristics advocate for existence of an array of quantum heterostructures with different quantization energies along the nanowire.

Finally, we present evidence that the observed emissions are indeed due to spatially indirect transitions from confined carriers in type II heterostructures. Spatially indirect transition exhibit longer life times due to the reduced overlap of the wave functions. We assessed this by performing time resolved measurements. In Fig. 3 the decay of the PL intensity with time from the emission at 1.48 eV is presented. Indeed, the observed life time is 8 ns.

In conclusion, we presented formation of high quality zinc-blende/wurtzite heterostructures in GaAs nanowires. This work opens a route towards band gap engineering using only one material.


Fig. 2: (a) PL spectrum from zinc-blende/wurtzite heterostructures in single GaAs nanowire. (b) Spatially resolved PL spectrum from the same nanowire. (c) Composite CL image containing three different emission energies. The different emissions are localized along the nanowire.

Fig. 3: Time resolved PL of the peak at 1.480 eV.
Growth of III–V semiconductor nanowires on Si (111) by MBE


The desire for monolithic integration of high–performance III–V compound semiconductors on silicon (Si) has fueled significant research over the years. In particular, the group–III-As materials (GaAs, AlAs, InAs and their compounds) are expected to play an important role, due to their excellent electro–optical properties for advanced high–speed electronic and near–infrared optoelectronic applications. Implementation of free–standing III–V semiconductor nanowires on Si substrates via the vapor–liquid–solid growth has recently emerged as a promising strategy to overcome critical lattice matching requirements with Si [e.g., Δa/a(InAs/Si) ~11.6 %], allowing very efficient strain accommodation. In addition, for functional nanowire devices on Si platform it is essential to have also control over position, geometry and size of the nanowires. Although most work so far concentrated on MOCVD growth of III-As-based nanowires either with or without the use of gold (Au) catalyst, very limited – and almost entirely Au–catalyzed – attempts to grow free–standing III-As nanowires on Si by MBE have been reported.

Since MBE offers several inherent advantages (low impurity content, sharp composition/doping control and the capacity for sophisticated core–shell heterostructures), our efforts are directed to establish controlled growth of free-standing vertical III-As based nanowires on Si (111) in a dedicated ultra–high purity MBE system. To allow self–induced growth (i.e., without the use of foreign catalysts such as Au) of well–oriented free–standing III–As nanowires on Si (111) substrates, two strategies are employed; (i) by use of an ultrathin SiOx mask on Si (111) to generate self–assembled nanowires and (ii) by means of prepatterned Si (111) substrates using electron beam lithography and etching techniques to facilitate site–selective growth of nanowires.

The scanning electron microscopy (SEM) images in Fig. 1 illustrate the two different approaches on the basis of InAs nanowires grown (a,b) by self–assembly and (c,d) by selective area epitaxy. The first two SEM images (Fig.a,b) are representative for self-assembled InAs nanowire growth on amorphous SiOx/Si(111) at temperatures of (a) 430 °C and (b) 505 °C and As-rich conditions (In/As flux ratio = 0.17). The InAs nanowires exhibit vertical directionality related to the {111}–family of ori-

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orientations, indicating direct epitaxial relationship to the Si (111) substrate. The majority of nanowires showed straight non–tapered morphologies with hexagon–shaped geometries. In addition, significant variation in nanowire length and diameter was found in the temperature range of 400–505 °C; however, growth of nanowires ceased at temperatures above 520 °C due to thermal instability.

On the other hand, prepatterned the Si(111) substrate by means of electron beam lithography and subsequent etching of a thin thermally grown SiO_2 nanomask [Fig. 1(c)] provided site-selective growth of InAs nanowires, shown here for a growth temperature of 480 °C at similar flux conditions (In/As = 0.17). The nucleation and growth of (111)–oriented InAs nanowires occurred preferentially at the predefined holes with very high yields of up to 80 percent. The size variation of the nanowires was found to depend critically on the pitch and growth time but much less on hole size at given growth conditions. Surprisingly, despite these size–dependent effects the nanowire geometries remained unaltered with uniform non–tapered morphologies. These findings suggest that surface diffusion kinetics, different adatom capture areas can be exploited to tailor precisely the size of III–V semiconductor nanowires in selective area growth methods.

The epitaxial relationship between the InAs nanowires and the Si (111) substrate was further confirmed on self–assembled InAs nanowires by HR–XRD 2θ–ω scans (not shown here). Over a wide 2θ range (0 – 60 deg) only the zincblende InAs (111) peak at ~25.3 ° and Si (111) peak at 28.35 ° were observed. More details on the microstructural quality of the as–grown InAs nanowires were extracted from TEM imaging (Fig. 2). The bright–field cross–section image in Fig. 2(a) shows a typical nanowire aligned vertically on the SiO_x/Si (111) substrate. Analysis of HR-TEM images [Fig. 2(b)] and selected area electron pattern [Fig. 2(d)] reveal that the nanowires have zincblende structure and grow on the (1-11) plane on top of the thin SiO_x layer.

Furthermore, low–temperature (20K) PL using an excitation laser wavelength of 808 nm yielded a peak emission at 0.445 eV on a typical nanowire array with a full width at half maximum (FWHM) of ~33 meV. It is worth noting that the PL peak emission of the InAs nanowires is slightly blue–shifted (30 meV) with respect to the bulk InAs reference, indicative of the existing radial quantum confinement effects.

In conclusion, we demonstrated self-induced (without external catalyst) growth of free-standing III–V nanowires grown by solid-source molecular beam epitaxy in both self–assembled and site–selective fashion. The latter is expected to be useful to allow good control not just over position, but also size and geometry, allowing the fabrication of well-defined nanowire based device structures.

Fig. 2: Cross–section TEM images of typical InAs nanowire nucleated on Si(111) presented in (a) low magnification, (b) high–resolution in the near center region and (c) at the bottom nanowire/substrate interface; (d) SAED pattern evidencing preferential [111] growth orientation. [Courtesy of Dr. J.-P. Zhang, Suzhou Institute of Nano-Tech and Nano-Bionics, China].

Supported by Marie Curie FP7 Reintegration Grant, DFG excellence program Nanosystems Initiative Munich and collaborative research center SFB 631
Thermoelectric figure of merit of nanocrystalline silicon measured by Harman method

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Nanostructuring has been shown to improve thermoelectric materials [1]. Two mechanisms related to nanostructure are relevant for increasing the conversion efficiency [2]. (i) Modulating the material composition can lead to a mismatch of acoustic phonon frequencies and consequently to a decrease of the thermal conductivity \(\kappa\). (ii) Structuring the material on the nanoscale can also lead to pronounced features in the electronic density of states due to lower dimensionality. Proper positioning of the Fermi energy thus may enhance the Seebeck coefficient \(S\) and the electrical conductivity \(\sigma\).

Material properties enter the device efficiency via the figure of merit \(ZT\), where \(T\) is the mean temperature and \(Z = S^2 \sigma / \kappa\) [3]. Measuring the parameters determining \(Z\) independently gives better insight into physical properties but is hampered by the inaccuracies of three different measurements. Harman developed a method to directly measure \(ZT\) by a current-dependent voltage measurement [4]. The method is based on the evolution of a temperature gradient along the sample due to the Peltier effect. For a constant current, the total voltage drop along the sample is the sum of the Seebeck voltage \(U_{\text{Seebeck}}\) and the voltage due to ohmic losses \(U_{\text{Ohm}}\). It can be shown that \(ZT = U_{\text{Seebeck}} / U_{\text{Ohm}}\). Usually, \(U_{\text{Seebeck}} + U_{\text{Ohm}}\) and \(U_{\text{Seebeck}}\) are measured in steady-state conditions and immediately after switching off the current, respectively.

Here we present Harman measurements on gas-phase grown and spark-plasma sintered nanocrystalline silicon, showing a typical grain size of 30 nm. Doping with nominally 5\times10^{20} \text{cm}^{-3} \text{P} leads to n-type conductivity. The sample is contacted with chromel wires and silver paste at the ends for current injection. In between, two type K thermocouples are placed for voltage (measured between chromel/chromel) and temperature (chromel/alumel) measurements. Figure 1 shows a typical measurement cycle. In the AC/DC mode (0 s – 50 s) \(U_{\text{AC}}\) can be measured more exactly, whereas in the transient mode (50 s – 100 s) time information can be resolved better. Besides \(U_{\text{Seebeck}} + U_{\text{Ohm}}\) and \(U_{\text{Ohm}}\), we also determine \(U_{\text{AC}}\) and \(U_{\text{DC}}\), where AC re-

\[\text{Fig. 3: Typical measurement cycle. The voltage (upper panel) evolves due to the current (lower panel) through the sample.}\]
fers to a square wave current and $U_{\text{DC}}$ equals $U_{\text{Seebeck}}+U_{\text{Ohm}}$. To obtain $ZT$, we use the relation $ZT = (U_{\text{DC}}-U_{\text{AC}})/U_{\text{AC}}$. Measuring $U_{\text{AC}}$ is advantageous since Joule heating of the sample is constant in the AC/DC mode and parasitic temperature gradients only appear as a voltage shift [5]. In the transient mode, where the current is simply switched on, a temperature-dependent conductivity can incorrectly influence the ratio $U_{\text{Seebeck}}/U_{\text{Ohm}}$ and therefore $ZT$. Due to limitations in measurement speed, the AC frequency in our measurement typically is 3.5 Hz, introducing the main inaccuracy measuring $U_{\text{AC}}$. The peak value of the square wave current is equal to the DC current and typically amounts to 0.5 mA.

Besides $ZT$, we can also deduce $S$ from a linear fit of $U_{\text{Seebeck}}$ vs. $\Delta T$, obtained from different current levels. Since we only can measure $\Delta T=T_{\text{hot}}-T_{\text{cold}}$ for positive and negative current polarity at one end, we assume that $T_{\text{hot}}-T_{\text{cold}}$ for one polarity. The electrical conductivity is given by $\sigma = I/\Delta U_{\text{AC}}l/A$, with $l/A$ accounting for the sample geometry. Combining $ZT$, $S$ and $\sigma$ we can calculate the thermal conductivity $\kappa = (S^2\sigma T)/ZT$. Figure 2 shows all four quantities relevant for thermoelectricity. As expected for silicon at this temperature range, $ZT$ rises continuously with temperature, starting from 0.018 at 300 K and reaching 0.15 at 700 K. $\sigma$ exhibits a slight drop at elevated temperatures. Though the uncertainties may be large, the calculated $\kappa$ does not show a decrease with temperature as known for crystalline Si [6]. Assuming negligible geometric effects of the contacts, the thermal conductivity $\kappa$ at 300 K is about one order of magnitude smaller compared to single-crystalline Si [6]. Room temperature measurements with a laser-flash method showed $\kappa \approx 11 \text{ W m}^{-1}\text{K}^{-1}$.

In conclusion, we demonstrated that a modified Harman method can successfully be applied to bulk semiconductor samples and all thermoelectric quantities can in principle be obtained. The investigated sample raises hope that the approach of nanostructuring also boosts the figure of merit for Si and the Si-Ge alloy system. Further systematic studies on a large number of samples can quickly be attained with this method.


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Supported by the DFG Schwerpunktprogramm SPP 1386 “Nanostrukturierte Thermoelektrika”
Defect reduction in silicon nanoparticles by low-temperature annealing

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Extensive research on freestanding crystalline silicon nanoparticles (Si-NPs) has been stimulated over the last years by the capability of producing sizable amounts of this material and its low-cost processing. Due to their favorable electronic properties, Si-NPs are of broad interest for the application in e.g. photovoltaics, thermoelectrics, lighting, and printable electronics [1]. However, previous studies have revealed that the conductivity of thin films of Si-NPs is limited by Si dangling bond (db) defects at the Si-NP surface which act as compensation and recombination centers. Therefore, one main focus of our work lies on an efficient defect reduction which improves the (opto)electronic properties.

The Si-NPs are grown in a low-pressure microwave plasma reactor by decomposition of silane. By adjusting the pressure and concentration of the process gases, the mean diameter of the Si-NPs can be tuned between 5 and 50 nm. With the help of electron paramagnetic resonance (EPR), we have investigated the change of the Si-db defect density after etching the particles with hydrofluoric acid (HF), after a vacuum annealing step at different temperatures from 150°C to 550°C as well as after a combined treatment. Fig. 1(a) shows typical EPR spectra for nominally undoped Si-NPs before and after annealing in vacuum at different temperatures. (b) [Si-db] as a function of the annealing temperature, in comparison to a H effusion spectrum (c). (d) Defect reduction for different particle sizes.

The dashed line results from a powder pattern of a defect with axial symmetry which can be attributed to trivalent Si atoms at the Si/SiO₂ interface, so-called P₃ centers. The symmetric signal at $g=2.0055$ (dotted line) has a signature which is typically observed for Si-dbs in a disordered environment which are referred to as D centers. In Fig. 1(b), the Si-db defect densities [Si-db], which were extracted from the EPR spectra by double numerical integration of the measured resonance peaks, are plotted as a function of the annealing temperature. We find that vacuum annealing of the Si-NPs at 200°C leads to a significant reduction

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of [Si-db] by a factor of ~8 with respect to the as-grown value. For higher annealing temperatures (350-500°C), we observe an increase of [Si-db] which can be understood by the desorption of hydrogen (Fig. 1(c)). Furthermore, a systematic investigation of the influence of the particle size on the ratio of [Si-db] before and after annealing at 200°C for 30 min is shown in Fig. 1(d). Within the experimental uncertainty, we do not observe a significant dependence on the Si-NPs size and [Si-db] is reduced by a factor of 5 in average.

Furthermore, the influence of HF etching on [Si-db] was investigated as it is a standard method to remove the native oxide from silicon and to obtain a H passivation of the surface. Figure 2(a) shows a typical EPR spectrum for etched Si-NPs with a mean diameter of 12 nm. The EPR data show that HF etching mainly removes the P₆ centers, which are directly located at the Si/SiO₂ interface. However, the density of D centers remains nearly unchanged, which suggests that they are not located at the outermost shell of the Si-NPs. In Fig. 2(b), we have also investigated samples where HF etching was combined with a thermal treatment at 200°C for 30 min. Both a preceding as well as a subsequent HF etching lead to a further significant defect reduction. In particular, if the native oxide shell is removed prior to the annealing, [Si-db] can be diminished by a factor of 10 compared to the as-grown value.

As EPR is only sensitive to neutral Si-db defects (paramagnetic), further experiments have been carried out to exclude that the annealing leads only to a redistribution of charges, rendering a fraction of the Si-dbs EPR inactive, rather than a true defect reduction. To confirm this, we have performed photothermal deflection spectroscopy measurements where we found a reduced subband gap absorption of etched Si-NPs before and after the annealing. Furthermore, photoconductivity measurements on thin Si-NPs films revealed that the annealing leads to an improvement of the photoconductivity by more than one order of magnitude.

In view of all the available experimental data, the low-temperature annealing of Si-NPs at 200°C under vacuum leads to a true reduction in the Si-db density. Our experiments demonstrate that this well reproducible process can be the basis for Si-NP layers with less trap states and higher conductivities which can make these NPs more attractive for application in a variety of electronic devices.

Metal-induced crystallization of silicon for thin film applications

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Polycrystalline silicon thin films on low cost substrates are attractive for large electronics and solar cell applications. One promising method to obtain large-grained high quality polycrystalline films is low-temperature crystallization of an amorphous precursor material by metal induced crystallization of silicon [1], especially the metal-induced layer exchange (MILE). Here a metal/amorphous Si layer stack, separated by a diffusion barrier, is annealed at temperatures up to the metal-silicon eutectic temperature, leading to an exchange of the position of the initial layers and the crystallization of the amorphous Si.

We have studied the MILE-process using aluminum (ALILE), silver (AgILE) [2] as well as different diffusion barriers to fabricate silicon thin films. The microscopic process of the MILE is independent of the used metal: During the annealing step, silicon atoms diffuse through the diffusion barrier into the metal layer, causing supersaturation. At a certain critical concentration of silicon atoms in the metal, the nucleation of silicon crystals is favorable. Until the concentration is reduced below the critical concentration by the silicon consumption of the growing grains, nucleation as well as grain growth is possible. Below the critical concentration no further nucleation is possible but grain growth proceeds. The initial metal layer is replaced by the growing silicon grains. Once all the grains are coalesced, the layer exchange is completed and the metal and the silicon layer have exchanged their positions and the amorphous silicon has crystallized. In this work, the different stages of the MILE are analyzed by energy dispersive X-ray analysis (EDX). At the beginning of the annealing step, a strict separation of the different layers is visible in the EDX-mapping (Si green, Al red, O blue (SiO₂-substrate, Fig. 1a). After the layer exchange, the different layers are separated again, but the metal and the silicon have exchanged their positions (Fig. 1b). The structural properties of the final polycrystalline layers were analyzed by Raman-spectroscopy and high resolution X-ray-diffraction. For both methods, ALILE and AgILE, the grown poly-Si layers are of high quality, but contain some stacking faults (not shown here).

Fig. 1: EDX-mapping images of an ALILE-layer stack (a) before and (b) after annealing at 450 °C.

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While the general process flow is the same for both crystallization methods, ALILE and AgILE, differences in the maximum applicable annealing temperature and the electronic properties are found. The maximal annealing temperature is strongly coupled to the eutectic temperature of the metal-semiconductor system (577 °C for Al-Si and 840 °C for Ag-Si). As crystal size and the process duration are increased at lower annealing temperatures, the suitable annealing temperatures are between 300 °C and 550 °C for the ALILE-process and 500 °C to 800 °C for the AgILE-process, leading to a layer exchange with crystal sizes up to several 10 µm and a duration of less than one day.

The electronic properties, especially the doping, of the poly-Si layers are strongly influenced by the solubility of the metal in silicon. Aluminum is an acceptor in Si with an activation energy of 54 - 72 meV and has a high solubility. This causes carrier concentrations up to $10^{19}$ cm$^{-3}$ in the poly-Si-layer. In contrast, silver has a solubility of $10^{13}$ cm$^{-3}$ in silicon and leads to intrinsic poly-Si.

In addition to the used metal also the diffusion barrier between the metal and the silicon has a strong influence on the process. In the ALILE-process, the natural oxidation of the aluminum layer is used as the diffusion barrier. For the AgILE an “artificial” diffusion barrier is of advantage due to the low tendency of silver to form an oxide.

We could show that a thin oxidized titanium layer can act as diffusion barrier in the AgILE as well in the ALILE. This modified metal-induced layer exchange is called titanium-assisted metal-induced layer exchange (Ti.MILE, EP2133907A1). One of the most important advantages of the Ti.MILE is the reduction of the crystallite density in the poly-Si. While a standard ALILE layer shows a crystallite density of $7 \times 10^5$ cm$^{-2}$ at 500 °C and $1.5 \times 10^5$ cm$^{-2}$ at 450 °C the crystallite density of a standard Ti.ALILE layer grown at similar conditions is $1.8 \times 10^5$ cm$^{-2}$ (500 °C) and $1.2 \times 10^4$ cm$^{-2}$ (450 °C) (Fig. 2). The corresponding crystal diameter is 25 µm to 100 µm for the Ti.ALILE.

We grateful thank Balaji Birajdar and Erdmann Spiecker from the Universität Erlangen for the support, the TEM- and EDX-analysis and the fruitful discussions. This work is supported by the “Fondsgesellschaft Dritte Patentportfolio Beteiligungsgesellschaft mbH & Co. KG”.


Random pn-junctions for physical cryptography

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Traditional cryptographic methods purely rely on mathematical assumptions and are vulnerable against increasingly powerful computers and the development of better breaking algorithms. For this reason physical cryptography receives increasing attention. Exploiting the inherent complexity and irreproducibility of physical (nanoscale) systems may provide fundamentally higher security than purely numerical schemes. One of the central tasks in this emerging field is to find systems that can be regarded as secure physical unclonable functions (PUFs). An ideal PUF (i) contains a very high amount of structural information, (ii) this information can be reliably extracted to create stable challenge-response pairs (CRPs). (iii) The rate at which the information is extractable, and the high number of all possible CRPs, prevent full characterization within short time by an adversary, and (iv) no computational model can numerically predict or imitate the PUF’s challenge response behavior.

We propose crossbars with fixed (non-writable), random information content as PUFs. The information content is stored in random diodes located at each crossing, determining challenge response pairs by their stable, but random, current-voltage characteristics. Slow read-out and very high information content are the key components of the proposed system [1].

Although the aluminum-induced layer exchange (ALILE) method [2] already has been used to prepare crossbar structures [1], here we focus on the fabrication of the individual diodes, which are produced by aluminum-induced crystallization on n-type Si-wafers. Hereby, silicon/Al/oxide/amorphous silicon (a-Si) layer structures are annealed at temperatures below the eutectic temperature of the Al-Si system (577°C), leading to a complete layer exchange and the crystallization of the a-Si [2]. After the annealing step, a closed polycrystalline silicon (poly-Si) layer is formed on the substrate, which is covered with a network of Al+Si-islands (hillocks). The inherently random nature of such a crystallization process makes it interesting for producing diodes with unique properties, since the actual crystallization sites cannot be predicted or controlled. Indeed, random I(V)-characteristics were found, with the diode characteristics strongly depending on the wafer doping, diode size, and poly-Si thickness.

N-type silicon wafers covered with a 100 nm thick thermal oxide (dark gray area, Fig. 1a) were used as a substrate for the crystallization process. To define the active area (10², 20², 100² μm²), the SiO₂ was structured by photo-lithography and removed by wet chemical etching in buffered HF solution (white area, Fig. 1a). After a second lithography step, the precursor layers for the ALILE process (Al/oxide/a-Si) were deposited. The photoresist was removed leaving behind the precursor layers only in the pre-structured area (light gray

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area, Fig. 1a). After that, the samples were annealed in dry N$_2$ atmosphere at 550°C until the layer exchange was completed. The poly-Si layer thickness was 20, 50, and 100 nm. A sketch of the diode structure after annealing is shown in Fig. 1b.

Electrical contacts to the diodes were made by the Al+Si-hillocks top layer formed after the layer exchange and Al evaporated on the backside of the Si wafers, resulting in a sandwich structure. Rectification ratios up to $2\times10^7$ measured at +/- 1V were obtained for diodes on a weakly doped wafer ($\rho = 1\text{ - }10\,\Omega\text{cm}$).

To identify possible design rules for the proposed crossbar structures, it is of importance to understand the origin of the randomness in our diode characteristics. Fig. 2a displays a histogram of $|\log I/A (+1V)|$ measured at 1 V for the 100$^2$ $\mu$m$^2$ -diodes on the weakly doped Si-wafer depending on the film thickness. While the randomness (spread of the histogram over the $|\log I/A (+1V)|$-axis) is comparable for the 50 nm and 100 nm thick films, it is considerably larger for $d_{\text{poly-Si}} = 20$ nm. This shows that a smaller film thickness leads to increasing inhomogeneity of the films, which results in the observed increase in diode randomness. It can also be seen that the average forward current is decreasing for smaller film thicknesses, which can be attributed to a higher series resistance for thinner poly-Si films.

In addition to the film thickness, the diode size has a considerable influence on the randomness of the I(V)-curves (Fig. 2b). For a poly-Si thickness of 100 nm, where only little randomness should stem from the film thickness itself, we find little randomness for the 100$^2$ $\mu$m$^2$ and 20$^2$ $\mu$m$^2$-diodes. Smaller diodes, on the other hand, exhibit randomness over more than two orders of magnitude. This can be attributed to a stronger fluctuation of the grain boundaries in the active area for smaller diodes, while for large diodes grain boundary effects are averaging out.

In conclusion, we found that poly-Si diodes prepared by ALILE on Si-wafers exhibit sufficient randomness and rectification for the use in the proposed crossbar structures. The diode size and film thickness have a considerable influence on the randomness of the resulting I(V)-curves. For more details see [3].


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Supported by the International Graduate School of Science and Engineering (IGSSE) of the Technische Universität München
Organic-inorganic heterojunction with P3HT and n-type 6H-SiC: Determination of the band alignment

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The exact band alignment in organic/inorganic semiconductor heterojunctions is generally influenced by a variety of properties and is difficult to predict. For the organic/inorganic heterojunction made of poly(3-hexylthiophene) (P3HT) and n-type 6H silicon carbide (6H-SiC), the band alignment is determined via current-voltage measurements. For this purpose a model equivalent circuit, combining a thermionic emission diode and space-charge limited current (SCLC) effects, is proposed which describes the behavior of the heterojunction very well. From the fitting parameters, the interface barrier height of 1.1 eV between the lowest unoccupied molecular orbital (LUMO) of P3HT and the Fermi level of 6H-SiC can be determined.

All 6H-SiC substrates were homogeneously nitrogen doped to $1 \times 10^{18}$ cm$^{-3}$, with resistivities of $\sim 1 \times 10^2$ $\Omega \cdot $cm. Ohmic contacts were established on the optically-polished backsides of the samples by evaporation of 50 nm of 80% Ni-20% Cr alloy and 50 nm of Au followed by annealing. Sample surfaces were prepared by solvent cleaning, plasma oxidation, and HF etching to remove the surface oxide. All further processing was performed in an Ar glove box (<1 ppm of O$_2$ and H$_2$O). Well defined junction areas, 0.5 mm$^2$, were lithographically produced on surfaces with photoresist. Immediately prior to formation of SiC-P3HT junctions, the exposed regions of the SiC surface were etched with HF vapor. The P3HT was dissolved to 2 wt.% in 1,2-dichlorobenzene and spin-coated on the 6H-SiC, resulting in a layer thickness of 150 nm within the lithographically defined holes. As ohmic frontside contacts to P3HT, 7 nm thick semi-transparent Au contacts were evaporated onto the surfaces. Dark and illuminated current-voltage ($IV$) measurements were performed.

![Graph](image)

**Fig. 1:** Dark current-voltage characteristics of a P3HT/6H-SiC heterojunction in a linear (a) and a semilogarithmic (b) plot. Besides the measured data (squares), the fit (red line) and the different contributions to the fit, the shunt resistance $R_{Sh}$ (green line), the diode (blue line) and the SCLC (cyan line) are plotted.

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under an Ar atmosphere. Temperature-dependent and spectrally-resolved measurements were performed in a liquid nitrogen-cooled cryostat under high vacuum conditions.

In Fig. 1 the dark current-voltage \((I-V)\) characteristics (squares) of the 6H-SiC/P3HT heterojunction is shown in a linear and semilogarithmic plot, including the fit (red line) and the single components of the fit. As the \(I-V\) characteristic in the semilogarithmic plot (Fig. 1(b)) at small voltages is symmetric, this is a strong indication for a shunt resistance \(R_{Sh}\) which governs the reverse direction of the diode. Therefore the reverse direction is fitted by a linear fit \(R_{Sh} = dU/dI = 5.2 \ \Omega \cdot \text{cm}^2\), which is in excellent agreement with the experimental data and shown as the green line in Fig. 1(b). After correction for \(R_{Sh}\), the \(I-V\) characteristics were fitted by a model of a series connection of a diode and a SCLC resistance, which is included as a nonlinear resistance. The fit is shown as the red line in Fig. 1 and exhibits also a very good agreement with the data. Furthermore, in Fig. 1 (b) the different contributions of the SCLC (cyan line) and the diode (blue line) to the fit are shown, under the assumption that the whole voltage drops either across the SCLC element or the diode.

The extracted parameters of the diode, which describe the forward direction in the voltage range of 0.6 - 0.9 V, before the SCLC dominates at higher voltages, are the ideality factor \(n = 2.9\) and the saturation current \(J_S = 3.5 \times 10^{-8} \ \text{mA/cm}^2\). By using the equation

\[
J_S = A^* T^2 \exp\left(-\frac{q\Phi_{bn}}{k_B T}\right)
\]

from the thermionic emission diode theory, a barrier height of \(\Phi_{bn} = 1.1 \ \text{eV}\) for the barrier in the forward direction can be extracted from the saturation current. For the calculation the effective Richardson constant \(A^* = 194.6 \ \text{A/cm}^2\cdot\text{K}^2\) for n-type 6H-SiC is used. The forward direction is defined by the injection of the electrons from the Fermi level into the lowest unoccupied molecular orbital (LUMO) of the P3HT, as the P3HT is undoped and an injection of holes into the VB of 6H-SiC hence is not an option. The calculated barrier is therefore the barrier for the thermionic emission of electrons from the 6H-SiC to the LUMO of the P3HT, as the P3HT is undoped and an injection of holes into the VB of 6H-SiC hence is not an option.

The calculated barrier is therefore the barrier for the thermionic emission of electrons from the 6H-SiC to the LUMO of the P3HT. In Fig. 2 the band alignment of P3HT and 6H-SiC is shown, including the barrier as determined from the current-voltage characteristics and a band bending of the 6H-SiC at the interface. The latter was determined by surface photovoltage measurements of the 6H-SiC surface. An upward band bending of 0.3-0.4 eV could be measured, which can be attributed to a Fermi level pinning at defect states at the surface.
Electrical passivation and chemical functionalization of SiC surfaces by halogenation

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Despite the increasing technological maturity of SiC, few practical ways of chemical and electronic passivation of its surfaces are available. Hydrogen termination, which is achieved on Si by simple HF etching, requires precisely-controlled high temperature processing of SiC under ultra-high purity molecular hydrogen [1,2]. Conversely, X-Ray Photoelectron Spectroscopy (XPS) and High Resolution Electron Energy Loss Spectroscopy (HREELS) studies have shown that HF treatment of SiC yields OH-terminated surfaces and high defect densities corresponding to approximately 400 meV of surface band bending [3-5]. Here we demonstrate a plasma processing method, which yields halogenated, electrically and chemically passivated 6H-SiC surfaces.

The chemical composition on the surfaces following plasma treatment was investigated by XPS measurements. Figure 1(a) shows XPS survey spectra of (0001) 6H-SiC after wet chemical cleaning and hydroxylation (red) and subsequent chlorine termination (blue). Wet-chemically cleaned surfaces show the presence of oxygen at ≈ 535 eV and no evidence of chlorine. After termination, we observe a significant decrease in the O1s core level signal, suggesting a majority substitution of surface-bound oxygen by chlorine. A double peak arises in the Cl2p core level which can be explained by spin-orbit splitting and suggests mono-chlorine termination of the surface (Fig. 1(b)). Additionally, Si2p core level spectra do not show any evidence for a remaining oxide following plasma treatment (not shown).

In order to determine the electronic properties of plasma treated surfaces, contact potential difference (CPD) and surface photovoltage (SPV) measurements were performed, showing that chlorine termination results in a change in surface band bending (CPD and SPV, Fig. 1 (c) and (d)). CPD measurements after plasma treatment reveal negative surface dipoles indicating the for-

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formation of polar bonds, which is consistent with the findings from XPS showing that chlorine atoms are present on the surface. Immediately after plasma treatment of (0001) SiC, the sample shows a CPD value of $\approx 0.5$ V which decreases to $\approx 0.3$ V within the first 20 hrs. This behavior can be explained by the formation of a water layer upon exposure to ambient atmosphere. However, SPV measurements show that chlorine termination leads to almost flatband conditions with a SPV of $\approx 10$ mV which is stable for more than 20 hrs, indicating a significantly higher stability of the termination compared to chlorinated silicon.

Covalent attachment of self-assembled organic molecules to the semiconductor surface provides an additional means of electrical surface passivation while simultaneously preventing re-oxidation, supplying linker sites for anchoring more complex biomolecules, and in the case of electrically insulating molecules, acting as a tunneling barrier for charge transport. Here we demonstrate a method to organically modify SiC surfaces without the formation of interfacial oxygen bridge bonds. Figure 2 shows XPS core level spectra of chlorinated (blue) and subsequently TFA-amine modified (green) SiC surfaces. The C1s core level spectra from TFA-amine modified surfaces exhibit additional, clearly resolved peaks at higher binding energies (insets in Fig.2 (a) and (b)). The peak at $\approx 285$ eV can be attributed to surface-bound hydrocarbons, whereas the additional peaks at $\approx 289$ eV and $\approx 293$ eV arise from the C=O and CF$_3$ groups of the attached molecules.

In conclusion, we have successfully demonstrated the termination of 6H-SiC with chlorine. This termination results in electrically and chemically passivated (0001) 6H-SiC surfaces. The surfaces exhibit flatband conditions, indicative of successful passivation of surface defects, such as dangling bonds. Furthermore, this technique allows for additional surface modification using alkene-based chemistry, as demonstrated by the successful UV-induced functionalization of chlorinated SiC with TFA-amines. There exist various additional routes for the functionalization of chlorinated surfaces, which render chlorinated SiC a promising substrate material for future bio-inorganic hybrid structures.

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Fig. 2: XPS C 1s (a) and F 1s (b) core level spectra of chlorine-terminated and organically-functionalized Si-polar 6H-SiC.

Optoelectronics of a submicron oxidized line in surface conductive single crystalline diamond


Diamond spurs scientific and technological interest because of its high thermal conductivity, a large dielectric breakdown, and a large bandgap of ~5.5 eV. Further functionality can be achieved by utilizing undoped single crystalline diamond with a hydrogen termination, which exhibits a surface conductivity in air. The surface conductivity originates from a two-dimensional hole gas that is formed due to an upward bending of the valence band at the hydrogenated surface. The extent of the hole gas is in the range of a few nm, depending on the quality of the hydrogen termination. In addition, hydrogenated diamond surfaces exhibit a negative electron affinity. The negative electron affinity becomes positive when the surface is oxidized. The change in electron affinity leads to a depletion of the hole gas. Patterning a thin oxidized line into hydrogenated areas of a diamond surface results in a lateral energy barrier for the two-dimensional hole gas. The corresponding band alignment has been recently verified by Kelvin force microscopy, and it is sketched in Fig. 1(a).

Here, we investigated the sub-bandgap optoelectronic phenomena induced by such a barrier in hydrogenated diamond at room temperature [1]. The experiments were performed using an electronic grade, chemical vapor deposition (CVD)-grown single-crystalline diamond with [100] orientation. The surface conductivity is created by exposing the surface to a hydrogen plasma. The submicron lines were defined in single-crystalline diamond by electron beam lithography in combination with an oxygen plasma treatment. We present results of devices with lines exhibiting a lithographic width of 70 nm and 1 µm. All measurements were carried out at room temperature in a vacuum of ~10^{-5} mbar.

Under laser excitation (1.24 eV < E_{\text{photon}} < 3.35 eV) the resulting current I_{\text{photo}} was measured with a lock-in technique. A typical spatial dependence of I_{\text{photo}} is depicted in Fig. 1(b), when the laser was scanned across an oxidized line with a lithographic width of 1 µm. We found a maximum of the optoelectronic response at the position of the oxidized line [triangle in Fig. 1(b)].

Fig. 1: (a) Schematic view of the lateral band alignment of an oxidized line in between two hydrogenated diamond surface areas. (b) Conductive photoresponse of the hole gas along the x-direction across such a barrier at room temperature. Triangle (diamond symbol) depicts the position of an oxidized line (a position next to it) with a lithographic width of 1 µm.

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Fig. 2 depicts a full map of $I_{\text{photo}}$ as a function of the coordinates $x$ and $y$ for the same line as in Fig. 1(b). Bias-dependent measurements demonstrate that $I_{\text{photo}}$ is symmetric with respect to zero $V_{sd}$ [triangles in Fig. 2(b)]. The bias-symmetry at any coordinate suggests that the optoelectronic signal was dominated by a photoconductance effect and not by a photovoltaic effect.

As oxidized diamond is known to exhibit surface defect states the optoelectronic observations can be interpreted as follows. After a sub-bandgap illumination, electrons get excited into traps within the bandgap [vertical arrow in Fig. 1(a)]. The trapped electrons affect the electrostatic potential landscape of the circuit such that they lower the barrier for holes centered at the position of the oxidized line. In turn, an increased optically induced conductance can be measured. Most importantly, such a photoconductance is determined by the occupation of the defect states.

This interpretation was supported by a superlinear dependence of $I_{\text{photo}}$ on $V_{sd}$ at large laser power $P_{\text{laser}}$ [circles in Fig. 2(b)], for the laser being focused onto the center of an oxidized line [triangle in Fig. 1(b)]. The superlinear behavior is characteristic of a space charge limited current known to appear in wide-bandgap materials with defect states available as traps for charge carriers.

The energy of the defect levels, which cause the photoconductive gain, is found to be at $(2.6 \pm 0.2) \text{ eV}$ above the valence band [Fig. 2(c)]. The value compares reasonably well with the ones reported in literature for defect states in oxidized diamond surfaces. The defect states are different to the ones induced by nitrogen impurities that lie 1.8 to 2.2 eV below the conduction band. We observed a typical response time of the presented diamond based circuits in the order of a few hundreds of milliseconds [Fig. 2(d)]. This is distinctive-ly shorter than the characteristic decay time of ~3 hours found for nitrogen defect states.

Growth and characterization of hetero- and homoepitaxial ZnO / Zn$_{1-x}$Mg$_x$O single quantum wells

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With its direct bandgap of 3.3 eV, ZnO is considered to be a promising material for the fabrication of bright blue and near UV light emitters. Its high exciton binding energy of 60 meV facilitates efficient radiative recombination of electrons and holes even at room temperature and above. Moreover, the availability of ZnO bulk crystals offers the possibility of homoepitaxial growth of ZnO-films with high structural quality and a low density of defects, particularly dislocations. Finally, by alloying with Mg the bandgap can be increased up to 4.2 eV and ZnO/ZnMgO quantum well (QW) structures can be fabricated in which exciton confinement further enhances the radiative recombination efficiency.

ZnO / Zn$_{1-x}$Mg$_x$O single quantum wells have been grown by plasma-assisted molecular beam epitaxy (PAMBE), both heteroepitaxially on c-plane sapphire and homoepitaxially on (000-1)-oriented bulk ZnO. For the heteroepitaxial samples, in order to fabricate abrupt and well-defined heterointerfaces, first a 300 nm thick ZnO template with a smooth surface was grown according to ref. [1].

A number of QWs were grown simultaneously on a ZnO template on c-plane sapphire as described above and a (000-1)-oriented ZnO bulk crystal from Crystec. All ZnO bulk substrates had been annealed beforehand at 1000 °C in O$_2$ atmosphere in order to remove polishing defects. The actual growth process was as follows: In a first step, ZnO with a nominal layer thickness of 35 - 40 nm was deposited, directly followed by a 20 nm thick Zn$_{1-x}$Mg$_x$O barrier and the ZnO well, whose thickness was varied between 1.1 nm and 10.4 nm. The QW was capped by the second Zn$_{1-x}$Mg$_x$O barrier with the same layer thickness of 20 nm as the first barrier. The resulting sample structure is illustrated in Figure 1 (a).

**Fig. 1:** (a) Sample structure of hetero- and homoepitaxial QWs. (b) RHEED pattern during homoepitaxial growth, featuring a 3x3 surface reconstruction. (c) Typical AFM image of a QW structure.
In the case of homoepitaxial growth, after a few minutes a 3x3 surface reconstruction evolved, which can be seen by the appearance of weak streaks between the main diffraction streaks of the [10-10] reflection high energy electron diffraction pattern (Fig. 1 (b)). The reconstruction pattern became weaker as growth continued and completely vanished after the Mg shutter was opened. AFM measurements revealed smooth surfaces with a RMS roughness in the range of 0.2 - 0.4 nm and clearly visible atomic steps for both homo- and heteroepitaxial samples. Exemplarily, an AFM image of a heteroepitaxial SQW grown at 540 °C is shown in Figure 1 (c).

Figure 2 shows the Photoluminescence (PL) spectra at T = 4.2 K of the samples under investigation. All samples exhibit sharp bound exciton emission lines in the region between 3.35 eV and 3.37 eV that originate from the ZnO template and the bulk ZnO substrate. At the high energy side a rather broad emission from the Zn_{1-x}Mg_xO barrier is observed. From its energy position the Mg concentration \( x \) is deduced according to [2]. The values obtained are \( x = 0.08 \) and \( x = 0.07 \) for the heteroepitaxial and homoepitaxial samples, respectively. The emission line marked with arrows is attributed to excitonic recombination in the QW. For wide wells it coincides with the ZnO emission and it blue shifts for decreasing well width as expected according to the quantum confinement. For the heteroepitaxial samples a strongly replicated emission line labeled with D appears in the range 3.31 eV to 3.33 eV. According to [3] this D-band can be assigned to acceptor-like defects confined to basal plane stacking faults. In contrast to that, The D-band is not observed for the homoepitaxial samples. Furthermore, temperature-dependent PL measurements reveal that the quenching of the emission intensity of homoepitaxial samples is less pronounced than for the heteroepitaxial samples. This further demonstrates the beneficial influence of homoepitaxial growth on the optical properties.

\[ x = 0.08 \]
\[ D \]
\[ x = 0.07 \]
\[ \text{PL intensity [arb. units]} \]
\[ \text{Energy [eV]} \]

Figure 2: PL spectra measured at \( T = 4.2 \) K of the samples under investigation. The QW emission is marked by arrows, the labels indicate the well width.

InGaN nanodisks embedded in GaN nanowires

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Nitride nanowires (NWs) have attracted a lot of interest in recent years due to their advantages compared to thin films. NWs are strain free single crystals which can be grown on different substrates, such as Si(111), Si(100), or sapphire, avoiding the formation of structural defects due to the lattice mismatch as in heteroepitaxial thin films. The inclusion of InGaN nanodisks (NDs) allows the formation of quantum wells inside GaN NWs. The nanoscale dimensions of these NDs lead to a quantum dot-like behavior with higher temperature stability of the optical emission compared to quantum wells in thin films. With a suitable In content in the NDs light emission in the visible wavelength regime can be achieved.

We have investigated the optical and structural properties of InGaN/GaN nanowire heterostructures with embedded InGaN nanodisks (NDs) grown by plasma-assisted molecular beam epitaxy (PAMBE). Here, low resistivity n-type Si(111) substrates were exposed to a nitrogen plasma at the growth temperature of 775°C for 10 min directly before GaN NW growth to create a 2 nm Si nitride layer. The growth parameters were selected to form self-assembled nanowires which grow perpendicular to the substrate surface along the wurtzite c-axis. Details on the growth can be found in Ref. [1]. The structure is shown schematically in Fig. 1. The GaN/InGaN nanowire heterostructures consist of a base part (length 400 nm) which either is composed only of GaN or a combination of GaN and InGaN regions. This part is directly followed by the quantum well (nanodisk) structure. InGaN nanodisks with different nominal thicknesses (1.3 nm to 4 nm) were formed between barriers of 5 nm to 7 nm GaN. The cap layer (25 nm) was also GaN. The resulting wire diameters are in the range of 25 nm to 50 nm. The main challenge for the growth of GaN/InGaN heterostructures is the large difference in growth temperatures between these materials, which is mainly caused by the different desorption temperatures of Ga and In. InGaN already decomposes at the optimum GaN nanowire growth temperature of 775°C. Therefore, the GaN base region was grown at $T_{\text{sub,1}} = 775^\circ\text{C}$, whereas for the nanodisk part (and the InGaN base region, if present) a reduced growth temperature $T_{\text{sub,2}}$ between 500°C and 660°C was chosen. The samples were analyzed by room temperature (RT) and low temperature (4K) photoluminescence. The as-grown samples were excited by the 244 nm line of a frequency-doubled Ar$^+$ ion laser. Due to the high density of nanowires (approx. 200 µm$^{-2}$) and the relatively large laser spot size, always an ensemble of NWs is probed. The comparison of samples with different base designs, MQW regions, and growth parameters allows the identification of the luminescence origin.

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We gratefully acknowledge financial support by the European Commission via the project DOTSENSE (STREP 224212).

Fig. 2: PL spectra recorded at 4K for three different samples with combined GaN/InGaN base grown with different In flux. The third sample has no NDs.

Fig. 3: Room temperature PL spectra for three samples with pure GaN base and different ND heights.

Fig. 2 shows PL spectra (T= 4K) of three different GaN/InGaN samples with combined GaN and InGaN base part grown at a substrate temperature of T_{sub}=590°C. At 3.41 eV and 3.35 eV emission from the GaN base is observed. The InGaN-related emission is located at 2.1 eV for the sample grown at an In effusion cell temperature of T_{In}=750°C, and in the range of 2.3 eV to 2.4 eV for the samples grown with an Indium flux reduced by a factor of 2 (T_{In}=720°C), proving that the peak position varies with the In content. The third sample is a reference sample with only the GaN + InGaN base (the ND part was omitted), showing that strong InGaN-related emission can be found also in samples without intentionally grown NDs. One possible explanation is the formation of In-rich clusters due to alloy fluctuations in the InGaN base. The peak position of the 3.1 eV emission shows no significant change with In concentration. Temperature dependent PL measurements show that this emission band is completely quenched at room temperature.

In Fig. 3 the room temperature PL emission for three samples with pure GaN base part and different ND heights is shown. When reducing the ND height the emission energy can be raised from 2.32 eV to 2.42 eV and the full width at half maximum (FWHM) is reduced from 420 meV to 240 meV. These structures exhibit a good temperature stability of the emission. For the 4 nm ND sample, the intensity of the ND peak at room temperature is still 8% of the 7K value.

Downscaling AlGaN/GaN HEMT structures

John D. Howgate, Sebastian Schoell, Sonja Matich, Ian D. Sharp, and Martin Stutzmann

In recent years, semiconductor nano-technology has been a focus area for many groups. A large effort has been placed on bottom up approaches, particularly in the case of controlled growth of semiconductor nano-wires. In this work we focus on the GaN/AlGaN materials system, for which there are already numerous accounts of nano-scale growth methods, including laser-assisted growth, pulsed metal organic chemical vapor deposition, and plasma-assisted molecular beam epitaxy [1]. An intensively studied feature of this materials system is the two dimensional electron gas (2DEG) which forms at the interface between GaN and AlGaN alloys due to a spontaneous polarization effect. This has enabled the development of high electron mobility transistor (HEMT) devices, which can be grown on SiC or sapphire substrates. These materials can be structured, usually by lithography and reactive ion etching (RIE), to be used for applications including biosensors, high power electrical modulation devices, and even X-ray dosimeters [2]. Naturally, research has focused on down-scaling these devices, and GaN/AlGaN nano-structures have been grown with a 2DEG present [3]. However, in this report we investigate the possibilities of reaching nano-scale devices by a top-down approach using various ion milling and etching techniques. The ultimate goal is to decrease the shutdown voltage of the devices by simultaneous perpendicular and lateral depletion of the 2DEG. This, in turn, should increase the transconductive sensitivity and allow the implementation of lateral depletion devices. Furthermore, we use HEMT structures with an additional GaN cap layer in order to make the devices more chemically inert. Indeed, HEMT devices with a GaN cap layers have been well studied and were shown to be bio-compatible and well suited for electrochemical and biological sensor applications. Furthermore, we have recently demonstrated photocatalytic processes at the interface between GaN and organic self-assembled monolayers [4].

For the experiments, MOCVD-grown HEMT wafers obtained from TopGaN and composed of 2 µm GaN/25 nm Al$_{0.2}$Ga$_{0.8}$N /2.6nm GaN cap layers were used. The resulting structures are characterized by a polarization induced 2DEG with a sheet carrier concentration ($n_{2DEG}$) of $8 \times 10^{12}$ cm$^{-2}$ and a mobility ($\mu$) of 1100 cm$^2$/Vs. From these wafers, arrays of 40 mesa structures, individually fitting on 5×5 mm$^2$ samples, were created with RIE then contacted with ohmic Ti/Al contacts. Each array contained 24 contacted lines and 16 transmission line method (TLM) structures. Together, each of the arrays covered channel lengths between 5 µm and 2 mm, and widths between 1 µm and 2 mm. For sub-micron tuning of the widths, the structures

![Fig. 1 Conductance vs. line width for reactive ion etched structures.](image-url)

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were additionally thinned by Focused Ion Beam (FIB) milling. The resistance and conductance of each structure was extracted from cyclic current voltage (IV) curves. The complete array of data was used to subtract bulk conduction and contact resistance effects.

As depicted in Fig 1, the results showed that the 2DEG starts to disappear due to sidewall damage from the RIE etch at a width of ~7 µm with total collapse at ~2 µm. The larger channel width data points have a linear behavior with an intercept at zero line-width which is in accordance with Ohm’s law.

To further analyze the disappearance of the 2DEG at small channel widths, FIB was used on the pre-characterized RIE-etched samples to mill gate widths between 0.5 and 16 µm. An optimization run with different FIB powers can be seen in Fig. 2(A). It was found that it was necessary to use the lowest settings of 1 s and 10 pA to be comparable with RIE. Fig. 2 (B, C) show a typical correlation between FIB and RIE.

A total of 4 sets of channel width dependent RIE and FIB data where compiled, focusing on the interval between 1 and 6 µm (not shown here). From these data an extrapolated effective channel width of 400 nm ± 200 nm could be achieved prior to 2DEG collapse.

In conclusion, we find that RIE and FIB are suitable for creation of micron scale devices. However, significant damage to the sidewalls of the mesa structures results in breakdown of the 2DEG which initiates at channel widths of ~7 µm, with total collapse at ~2 µm. However, additional methods including post-processing thermal annealing, chemical etching techniques, and “softer” conditions that cause less damage are being investigated to further downscale device dimensions. In the future, these devices will be characterized and used in electrochemical and biological environments to study the application advantages of lateral depletion devices.

Catalytic nanoparticles on wide-bandgap semiconductor surfaces

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Wide bandgap semiconductors like GaN have been utilized for a myriad of applications. For those related to surface properties, like bio-functionalization or catalytical activity, it is interesting to understand the charge transfer mechanism over the semiconductor surface onto biomolecules or metal particles. In this project, platinum nanoparticles are applied to wide-bandgap semiconductor surfaces (GaN, diamond) with a defined distribution to create a catalytically active support. This is a prerequisite for the electrical or optical manipulation of the platinum catalyst via the semiconductor and hence a precondition for catalysis on demand.

To investigate the targeted charge transfer mechanism, Pt nanoparticles are deposited on the semiconductor surfaces via physical vapour deposition or spin-coating (Fig.1). To avoid agglomeration during spin-coating, the nanoparticles are protected by a Poly(N-vinyl-2-pyrrolidone) shell. In combination, these methods allow for a defined variation in coverage ranging from below 10\% of a monolayer to multilayers. However, regarding the precise control of catalysis kinetics as well as the total noble metal consumption, a low coverage is preferred. The deposited particles are stable under nitrogen flux, organic solvents and mechanical stress \cite{1}.

Temperature programmed reaction combined with \textit{in situ} grazing incidence small-angle X-ray scattering and X-ray absorption spectroscopy revealed the catalysts to be sintering resistant at elevated temperatures as well as during reduction and hydrogenation reactions \cite{2}. In contrast to Pt particles of approximately 7 nm diameter, smaller particles of 1.8 nm in size are found to dynamically adapt their shape and oxidation state to the changes in the reaction environment \cite{2}.

The electronic properties of the Pt-GaN interface can for example be characterized by photoelectron spectroscopy (XPS) (Fig.2). The characteristic Pt4f doublet is measured to determine the oxidation/reduction state of the Pt nanoparticles for small (1.8nm) and large (7nm) particles for n- and p-type GaN substrates. All samples exhibited two Pt doublets, indicating that a large fraction of the nanoparticles is not in its bulk native state but oxi-

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Funded by the International Graduate School for Science and Engineering (IGSSE), Project 2-3
dized. A core level peak shift of 0.3 eV was observed on p-type GaN, indicating charge transfer from the Pt Npts to the substrate (Fig.2c). The shift was increased to 0.8eV, if the GaN surface had been pretreated with aqua regia, which is known to remove GaOx from the surface (Fig.2d). For future excitation experiments, a micro-reactor setup with a volume of 50µl was built, and first activity tests were performed (Fig.3). The measured turnover frequencies (TOFs) matched the literature values nicely. An increase of the TOFs for increased temperature and initial activation time was observed.

In conclusion, the preparation of catalytically active compounds from Pt nanoparticles and GaN surfaces was presented. First results prove catalytic activity of the deposited nanoparticles, as well as a possible charge transfer to the GaN substrate. The future task will be the increase of efficiency and precision of catalytic reactions by optical or electrical excitation of the semiconductor substrates in the developed micro-reactor setup.

Fig. 2: XPS characterization of Pt Npts on n-type GaN (a-b) and p-type GaN (c-d), with and without aqua regia pretreatment. Core level peak shifts due to charge transfer are indicated by arrows.

Fig. 3: Pt catalyst activity tested in a micro-reactor setup. The turnover frequencies (TOF) are dependent on temperature as well as the initial activation time (reduction under $H_2$ flux).


Funded by the International Graduate School for Science and Engineering (IGSSE), Project 2-3
Field-enhancement in metal nanoparticle based circuits

Markus A. Mangold, Christoph Weiss, M. Calame, and Alexander Holleitner

The excitation of surface plasmons in two-dimensional arrays of metal nanoparticles (NPs) can lead to a strong enhancement of the electromagnetic field at the surface of the particles at visible and near-infrared wavelengths. In the context of molecular electronics, densely packed two-dimensional NP arrays have been exploited to electrically contact molecules. However, an open question remains whether surface plasmon resonances can be exploited to induce charge transport across nanoscale metal junctions with molecules embedded. To this end, we experimentally investigated the photoconductance properties of well ordered two-dimensional gold NP arrays consisting of millions of metal junctions, in which alkane molecules are incorporated.\(^1\)

Starting point were two-dimensional arrays of alkane coated gold NPs. After self-assembly at an air-water interface the arrays were patterned in 25 μm wide stripes on a SiO\(_2\) Chip by a microcontact printing technique. In a shadow mask evaporation step, these stripes were contacted by macroscopic gold electrodes with a separation of \(\sim 8\) μm, leaving a contacted area with dimensions of \(25 \times 8\) μm\(^2\). The scanning electron microscope (SEM) image in Fig. 1(a) shows a high resolution image of such a NP array. Predominantly, the NPs are arranged in a hexagonal array with a lattice constant of approximately 12 nm and a particle-particle distance of about 2 - 3 nm. Fig. 1(b) depicts the optically induced current \(I_{\text{PHOTO}}\) measured on a NP array as a function of laboratory time. The (red) highlighted areas indicate times where the laser light is on. (c) Photoconductance as a function of the laser spot position. Source and drain electrode are drawn schematically as a guide to the eye. (d) SEM-graph of a 25 μm wide stripe of NP array, contacted by gold electrodes with a distance of 8 μm.

![Fig. 1](image)

**Fig. 1:** (a) Scanning electron microscope (SEM) image of an array from gold nanoparticles (NPs) coated with octane thiols. (b) Optically induced current as a function of laboratory time. The (red) highlighted areas indicate times where the laser light is on. (c) Photoconductance as a function of the laser spot position. Source and drain electrode are drawn schematically as a guide to the eye. (d) SEM-graph of a 25 μm wide stripe of NP array, contacted by gold electrodes with a distance of 8 μm.

depends linearly on \(V_{\text{SD}}\) and we detect no finite value of \(I_{\text{PHOTO}}\) at zero bias. Thus, the measured signal is due to an optically induced change of conductance, and we can use the term “photoconductance”, i.e. \(G_{\text{PHOTO}} = I_{\text{PHOTO}} / V_{\text{SD}}\), to describe the observed phenomen-
non. In Fig. 1(c), the spatially resolved photoconductance of the area shown in Fig. 1(d) is depicted.

In Fig. 2(a) we show the photoconductance for different excitation wavelengths. We observe a maximum of the photoconductance at approximately 600 nm, which coincides with the surface plasmon resonance of two-dimensional arrays made out of octane coated NPs. Therefore, the strong maximum of the photoconductance at that wavelength suggests that surface plasmons play an important role in the creation of the photoconductance.

In order to describe our observations, we used the Maxwell-Garnett effective medium theory to calculate the absorption of the nanoparticle array. Details of the calculation can be found in Ref. [1]. In Fig. 2(b) the calculated absorption is plotted as a function of wavelength. We obtain an absorption spectrum with a maximum absorption at a wavelength of 595 nm, which is consistent with recent measurements. Hereby, the calculation corroborates the interpretation that the photoconductance maximum at 600 nm results from a surface plasmon resonance in the two-dimensional gold NP arrays. Under the assumption, that the heat transport from the nanoparticle array is limited by the underlying SiO$_2$ layer, we can directly calculate the expected bolometrically induced conductance from the absorption spectrum and the measured temperature dependence of the nanoparticle array’s conduction. We determined the photoconductance value at 600 nm to be $G_{\text{PHOTO}} = 81$ fS. The latter value is in the same order of magnitude as the experimentally determined photoconductance signal of 200 fS at 600 nm (Fig. 2(a)).

In conclusion, we report on a photoconductive gain effect in two-dimensional arrays of gold nanoparticles (NPs), in which alkane molecules are inserted. The NP arrays are formed by a self-assembly process from alkanethiol-coated gold NPs, and subsequently they are patterned on a Si/SiO$_2$ chip by a microcontact printing technique. We find that the photoconductance of the arrays is strongly enhanced at the frequency of the surface plasmon of the NPs. We interpret the observation as a bolometric enhancement of the conductance of the NP arrays upon excitation of the surface plasmon resonance.


Imaging of metallic nanoparticles on glass and gold surfaces

Alexander Schwemer, Markus Schuster, and Ulrich Rant

Metal nanoparticles are materials with exceptional optical properties: when excited at the appropriate frequency, the conduction electrons in the nanoparticles oscillate resonantly with the incoming light wave. In this state, termed as Localized Surface Plasmon Resonance (LSPR), light is strongly absorbed and scattered by the nanoparticles, making the detection of individual particles possible\cite{Wang2009}. Moreover, the LSPR wavelength depends on the nanoparticles’ local surroundings, which makes metallic nanoparticles nano-scale environmental sensors\cite{Kuzum2017}. Because of their dimensions are comparable to important biological macromolecules like DNA or proteins, nanoparticles are being successfully employed as nano-reporters in single-molecule biophysical and biochemical studies\cite{Li2015}. Single molecule and single particle experiments, however, require optical instruments with high sensitivity and high resolution.

Here we present an optical setup with a microfluidic sample chamber for the detection of single metal nanoparticles on dielectric (glass) and metallic (gold) surfaces. We utilize an inverted microscope (Olympus, IX71), schematically depicted in Fig. 1A. A darkfield (DF) condenser focuses white light at a wide angle onto the sample, which is located at the bottom of the flow-cell. The flow-cell was especially developed to feature a very low profile and consists of two microscope coverslips, assembled with an adhesive

\begin{figure}[h]
\centering
\includegraphics[width=\textwidth]{fig1.png}
\caption{(A) Scheme of an inverted microscope in darkfield (DF) configuration and a flow-cell with electrical contacts. A 100x objective collects scattered light from the sample, while rejecting excitation light. (B) CCD-chip for true color images. The image shows single 40 nm gold nanoparticles. (C) high-speed, electron multiplying CCD with an optical setup for simultaneous imaging at two spectral ranges. (D) Scanning spectrometer with photo multiplier tube (PMT). (E) Scattering spectra of single silver and single gold nanospheres (40 nm) plotted in blue and red, respectively.}
\end{figure}

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tape which is custom-cut to contain the flow channel. Thin gold films (thickness 20 nm) with an optical transmittance of 30% in the visible range are evaporated onto the lower coverslip. The flow-chamber is connected to a pumping system with Teflon tubes to study surface functionalization in situ. A 100x objective collects light scattered from the sample, while rejecting the excitation light. A color CCD camera (Nikon, D700) is connected to one exit port of the microscope to record true color images, see Fig. 1B and Fig. 2. It can be replaced by a high-speed, electron multiplying CCD chip (Hamamatsu, ImagEM), for high speed monochrome imaging of low light level samples (Fig. 1C). A special beamsplitter setup has been constructed which enables the high-speed simultaneous detection of the same region-of-interest at two different wavelengths (Fig. 1C). Detailed spectroscopic information is obtained by using the other exit port of the microscope which is connected to a scanning spectrometer (Jobin Yvon, Triax 550) with a photo multiplying tube (Fig. 1D). Fig. 1E shows normalized scattering spectra of single nanospheres with a diameter of 40 nm: A silver sphere peaks at 400nm (blue) and a gold sphere at 550nm (red). Fig. 2 shows a set of commercially available silver and gold nanoparticles with varying diameters; particles down to 10 nm in diameter are detectable in the flowcell.

**Fig. 2: RGB - image set of metal nanospheres differing in diameter, deposited onto a glass surface inside the flow-cell. The scale is valid for all images.**

The flowcell is designed to study the chemical interaction of single nanoparticles and a metallic surface. Fig. 3 shows the adsorption of PEGylated gold nanospheres onto differently coated gold surfaces: We find that the particles strongly adsorb onto gold surfaces which have been functionalized with mercaptohexanol (MCH, HS-(CH₂)₆-OH) self-assembled monolayers (SAM). Coating the gold with PEGylated SAMs (HS-(CH₂)₂-O-((CH₂)₂O)₆-CH₃) passivates the surface in comparison to a bare gold surface and suppresses non-specific adsorption of particles.

Quantum dot-carbon nanotube hybrids

B. Zeblia), H. Vieyraa), I. Carmelib), A. Hartschuhc), J. P. Kotthausa), and A. W. Holleitner1)

Carbon nanotubes (CNTs) have emerged as promising one-dimensional building blocks of nanoscale optoelectronic devices. Hybrid systems based on functionalized CNTs in particular hold interesting prospects in various fields such as optical sensing and photovoltaics. In this work, we functionalized CNTs with CdTe nanocrystals via molecular recognition.[1] The resulting hybrid systems combine the adjustable optical properties of quantum dots with the exceptional transport characteristics of nanotubes. Optical spectra of the hybrids were found to consist of a superposition of the quantum dot photoluminescence and Raman scattering from the carbon nanotubes, a unique signature used to retrieve structural information.[2] We showed that the photoconductance of the hybrids can be adjusted by the absorption characteristics of the nanocrystals. Surprisingly, the photoconductance of the hybrids exhibits a slow time constant of about 1 ms after excitation of the nanocrystals. The data are consistent with a bolometrically induced current increase in the CNTs caused by photon absorption in the nanocrystals. Our results contribute to our understanding of hybrids and will help in designing novel functional nanosystems.


Device Physics
Electrically pumped room temperature CW VCSELs for emission around 2.6 µm

S. Arafin¹, A. Bachmann, K. Vizbaras, and M.-C. Amann

Mid-infrared semiconductor lasers are key components for tunable diode laser absorption spectroscopy (TDLAS) based trace gas sensing applications. Among different semiconductor lasers, Vertical-Cavity Surface-Emitting Lasers (VCSELs) are very attractive choices for these applications because VCSELs exhibit excellent lasing characteristics such as a true single-mode operation, a high current tuning coefficient and a cost-effective production. Since most of the trace gases have strong absorption lines above 2 µm, the GaSb-based material system is used for covering this wavelength regime. Electrically-pumped (EP) VCSELs at 2.3 µm operating in pulse or continuous wave (CW) mode have already been reported [1,2].

Here we describe GaSb-based CW-operated EP VCSELs incorporating a buried tunnel junction (BTJ) with a record emission wavelength around 2.6 µm. Owing to the excellent thermal heat management, the devices exhibit single-mode operation beyond room temperature. In addition, the devices show features such as efficient (electro-)thermal wavelength tuning which can be utilized in gas sensing since several absorption lines of a single gas species can be scanned using the full tuning range of such lasers. The emergence of room-temperature laser at a wavelength of 2.6 µm promises significant sensitivity improvements for water absorption spectrometers owing to the approximately 20 times larger absorption strengths compared to 1.4 µm lines. In fact, the 2.6-2.65 µm wavelength (λ) region shows strong water vapor absorption lines making this range well adapted to water measurements and isotopic ratio studies. This wavelength regime especially at λ = 2.605, 2.596 and 2.594 µm are very important for the detection of water vapor since the strongest water vapor absorption lines are found exactly at these lines. Impressive results are obtained on VCSELs at above-mentioned emission wavelengths with sufficient optical output power, being well-suited for the targeted applications.

A schematic cross-section of the 2.6 µm GaSb-based VCSEL design is illustrated in Fig. 1. Utilizing a twofold epitaxial growth process, the VCSEL structure was grown with a Varian Mod Gen II-MBE system equipped with solid sources and valved cracker cells for arsenic and antimony. Details of the GaSb-based BTJ-VCSEL design and fabrication process can be found elsewhere [2].

Fig. 1: Schematic cross-sectional view of the 2.6 µm GaSb-based BTJ-VCSEL structure.

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Supported by the European Union via NEMIS (contract no. FP6-2005-IST-5-031845), the German Federal Ministry of Education and Research via NOSE (contract no. 13N8772) and the German Excellence Cluster “Nanosystems Initiative Munich (NIM)

Devices were tested on-wafer under cw operation on a temperature controlled heat sink. Fig. 2 (a) illustrates the temperature dependent L-I (light output-current) characteristics of a VCSEL with 6 μm BTJ diameter. Cw operation has been achieved up to 50°C. At RT, the threshold current of the device is 3.8 mA, corresponding to an effective threshold current density of 7.5 kA/cm², where lateral carrier diffusion of 2 μm has been taken into account. The maximum output power at 20°C is 176 μW @ 13.5 mA. Fig. 2 (b) shows the threshold current and maximum output power as a function of the heat sink temperature where it is clear that the minimum threshold current is located at a temperature below 20°C, hinting that the spectral alignment of the cavity mode and gain spectrum is not optimum even at this low temperature.

The emission spectra by varying driving currents at a constant heat sink temperature (T_HS) of 10°C are displayed in Fig. 3, yielding distinct single-mode emission. They show a side-mode suppression ratio (SMSR) of over 25 dB. The wavelength shifts at a rate of 0.5 nm/mA.


High-speed short-cavity VCSELs with improved modulation performance and high contrast grating

M. Müller¹, D. Schnurbusch, and M. Horn

With the emerging 100G-Ethernet standard and the advent of FTTX-solutions, cheap laser-diodes featuring high modulation-bandwidths, reasonable output-power, and emission in the near-infrared are required. In order to provide light sources for typical transmission ranges of access networks, great efforts in developing high-speed Vertical Cavity Surface Emitting Lasers (VCSELs) for the 1.3 μm and 1.55 μm wavebands have been made - with steady improvement [1-2]. Particularly for VCSELs with buried tunnel junction (BTJ) and novel short-cavity (SC) design, promising results and high modulation bandwidths have been shown [1].

The work at hand discusses the modulation performance of VCSELs with applied SC-design. Moreover, we introduce a novel type of out-coupling mirror based on a High-Contrast Grating (HCG) featuring polarization dependent reflectivity.

Fig. 1. (a) Schematic cross section of short-cavity VCSEL (b) L-I-characteristics at 25°C and 80°C; optical spectrum at 25°C and IDC = 14mA (c) Small-signal modulation response for various currents at room-temperature. The maximum achieved 3dB-cutoff frequency of 17 GHz is indicated for a current IDC = 12.4 mA.

1. Short-Cavity VCSEL: Design and Static Performance

Fig. 1.a) shows a sketch of the SC-VCSEL layout. The application of two dielectric mirrors with a high refractive index contrast among individual layer pairs, leads to a shorter effective cavity length, reduced photon lifetime, higher relaxation resonance frequency, and reduced intrinsic damping as shown in earlier work [3]. For reduced contact-pad parasitics, benzocyclobuthene (BCB) is used as passivation. Introducing compressive strain to the quantum wells, reduces the transparency carrier density, enhances the (differential) gain and, therefore, high-speed operation. The doping-level of the InP-regrowth layer has been lowered in order to reduce the parasitic capacitance of the space charge region, which strongly reduces the device parasitics [1].

The maximum optical output power exceeds 2 mW at room-temperature and 0.8 mW at 80°C for a device with an active diameter of 5 μm (Fig. 1.b). It should be remarked that the thermal roll-over is shifted to higher currents under large-signal modulation due to the reduced dissipated heat. Threshold currents are as low as 1 mA and 2 mA, respectively. Due

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to the high coupling efficiencies of up to 60% for these devices, high fiber-coupled power amplitudes are achieved. The spectrum shows single-mode operation with a side mode suppression ratio of more than 40 dB at roll-over current.

2. **Small- and Large-Signal Modulation Performance**

Figure 1.c) plots the S21-parameter over modulation frequency for room-temperature operation. Excellent device-parasitics is revealed by the pronounced resonance peaks observed for the low-current regime where intrinsic and thermal damping is negligible. The parasitic cutoff-frequency was determined to be as high as 23 GHz [3]. A maximum 3dB-cutoff frequency of 17 GHz is achieved for bias-currents close to thermal roll-over, where the resonance gets strongly damped due to increased thermal and intrinsic damping.

3. **High Contrast Gratings**

For many applications, polarization stability has been a challenge for VCSELs, as the structure in Fig. 1.a) does not have an intrinsic polarization preference. Novel high contrast gratings (HCGs) breaking the polarization degeneracy and providing high reflectivity at the same time, can therefore substitute typical DBRs in VCSELs. Fig. 2.a) shows a schematic layout of such a HCG consisting of silicon grating bars with high refractive index surrounded by air and SiO$_2$ with low refractive index. For appropriate grating geometry high reflectivities can be achieved for the TE-mode as the simulation results in Fig. 2.b) reveal. At the same time the reflectivity of the TM-mode is below 3%. Including optical losses in amorphes silicon, the TE-reflectivity exceeds the reflectivity of a typical DBR (dashed line) in a 50 nm wavelength-range around 1.55 µm. Therefore, the top-DBR of the VCSEL presented in Fig. 1 can be substituted by such a polarization selective HCG. Fig. 2.c) shows a SEM-image of the same fabricated HCG.

![Fig. 2. (a) Sketch of a Si/SiO$_2$-HCG (b) Simulated TE-reflectivity for a HCG with adequate grating geometry including optical losses in amorphes silicon (c) SEM-image of fabricated HCG.](image)

4. **Conclusion**

In this paper we presented our latest 1.55 µm High-Speed Short-Cavity VCSELs with record-high modulation bandwidth. For polarization-stability, we suggested the substitution of the top-DBR by a high contrast grating.

Simultaneous two-level lasing in GaInAsSb/GaSb strained quantum-well lasers

Kristijonas Vizbaras¹, Kaveh Kashani-Shirazi, and Markus-Christian Amann

Simultaneous two-state lasing has already been observed in quantum dot lasers [1], where it was attributed to a ‘phonon-bottleneck’ effect resulting in an increased carrier relaxation time between the excited state and the ground state due to combined effects of limited density of states in quantum dots and inefficient relaxation by phonon emission when inter-level energies do not match energies of the optical phonons. Such an inhomogeneous broadening by retarded relaxation in quantum dots has been also predicted theoretically [2]. Slow inter-subband relaxation in quantum wells has also been observed previously in asymmetric and coupled quantum well systems [3], however, up to now, it has never been observed experimentally to result in simultaneous two-level lasing.

We present first experimental evidence of simultaneous two-level lasing in quantum well lasers (QW) under continuous wave (CW) operation, at room temperature [4]. We attribute the excited level lasing to a slow interaction between the subbands. From spectral measurements it was observed that devices with longer resonator lengths (>400 µm) started lasing with an emission wavelength of 2720 nm, corresponding to the ground level transition of 0.456 eV. On the other hand, shorter devices started lasing at 2490 nm, corresponding to the second QW level transition of 0.5 eV. The latter is due to the fact that mirror loss is directly proportional to the inverse cavity length, therefore, by changing the cavity length the total loss is also changed and, at some point, gain in the ground level becomes smaller than the total loss and lasing is prohibited.

![Fig. 1: Measured emission spectra of 500µm long device, with a 30µm wide ridge, at a current density of 1.3 kA/cm², at 288 K, under CW operation. The laser line at 0.456 eV corresponds to the ground level transition, whereas the laser line at 0.5 eV corresponds to the excited level transition.](image)

![Fig. 2: Wavelength switching as a function of laser current for a 500 µm long device, with a 30 µm wide ridge.](image)

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A more detailed investigation was carried out on 500 µm long devices, which showed simultaneous lasing from both QW levels under CW operation (Fig. 1). It was observed that the wavelength switching from the ground level to the excited one has ‘digital’ character (Fig. 2) and that simultaneous lasing takes place only in a limited range of current densities. At low current densities (below 1.22 kA/cm²) only ground level lasing is observed. With increasing current second level lasing appears as well and simultaneous lasing from both levels takes place. Increasing the current further, the ground level transition ceases, and only lasing at the second level transition is observed (above 1.32 kA/cm²) (see Fig. 2).

Our experimental observations indicate that no filling takes place. The population of the ground level gets clamped immediately after the threshold is reached. This can be seen from a nearly constant laser emission wavelength with increasing current (Fig. 2). The appearance of the excited level lasing indicates that the population in the second level does not clamp when the lasing condition is reached in the ground level, what is known to be in the case of finite inter-subband relaxation time [5]. Such filling, eventually, leads to population inversion in the second level as well. The observed behavior indicates that very slow scattering between the subbands exists. The exact origin of such a behavior is not clear. Typical longitudinal phonon (LO) energies in GaInAsSb are around 25 – 30 meV and are smaller than inter-subband spacings. Polar electron-phonon interaction is dominated by longitudinal optical phonon scattering and is inversely proportional to the square of phonon momentum, thus in GaInAsSb it is expected to be weaker, because phonons with larger momentum need to be involved. With the build-up of carriers in the excited state, screening of electron-phonon interaction by free carriers is expected to play an important role, especially at carrier densities higher than $10^{17}$ cm⁻³ [6].

To summarize, we have demonstrated simultaneous lasing from two QW laser levels under continuous wave operation. We attribute the effect to the slowed-down inter-subband as no band filling was observed. Such behavior indicates strong carrier accumulation in upper laser levels, what is known to deteriorate greatly internal efficiency of semiconductor lasers with quantum confined active regions [5], and, therefore, could give an explanation for poor internal efficiency typically observed for GaSb-based lasers.

Empirical modeling of the refractive index for (AlGaIn)As lattice matched to InP

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The AlGaInAs/InP alloy has become the standard material system for long wavelength laser diodes like vertical-cavity surface-emitting lasers (VCSELs), and high-speed devices usable for 100-G Ethernet have been fabricated [1]. The improved performance compared to GaInAsP-based laser diodes is mainly ascribed to the higher conduction band offset and therefore reduced electron leakage current from active regions. Since the minimization of free carrier losses and the positioning of the tunnel junction, which is needed as current aperture, into a node of the optical field is essential for optimal performance and low threshold currents for such devices, precise knowledge of the corresponding refraction indices is extremely important for the design of the layer structure and doping profile. A composition dependent refractive index of (AlGaIn)As including dispersion was derived by reflection spectroscopy using a new fitting procedure for the whole measured data range, yielding an up to now unmatched accuracy. These data are compared with the reported values from Mondry et al. [2] and Nojima et al. [3], who used a similar technique. Significant deviations, which are critical for design considerations, could be revealed.

To measure the refractive index we used a reflectance technique already described in [2]. It is based on a Fabry-Pérot etalon structure consisting of an epitaxial layer of unknown refractive index, which is in our case AlGaInAs layer with 2 µm thickness and different compositions, on a substrate, here InP. The spectra have been measured with a Bruker Vertex V70 FTIR spectrometer. The AlGaInAs layers were grown by MBE as well as MOVPE.

The reflection coefficient of this structure depends only on the field reflectivity at the epitaxial layer/air interface, the reflection at the epitaxial layer/substrate interface, the thickness of the AlGaInAs layer and the corresponding k wavevector. Since the reflection coefficient of InP is well known, only the refractive index and the thickness of the layer were released as parameters by using a least square error fit method. The unknown refractive

Fig. 1: The normalized etalon-reflection (straight lines) extracted from the measurements is illustrated. The resulting fit is indicated as dashed line.

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index of the AlGaNAs epilayer was estimated by using a single oscillator Sellmeier equation. The AlGaNAs fitting data of the grown samples have been used to create a third order interpolation matrix with respect to “effective” composition $X$ (see description of Fig. 2) for the refraction index of this material system lattice matched to InP. Fig. 2 compares the index dispersion of (AlGaN)As with various fractions $X$. A comparison with grown AlInAs/GaInAs DBR structures confirmed our findings, since the bandwidth of the DBR is given by the refractive index difference of the grown layers, which is only for our data sufficient high (see Fig. 2). Finally, it should also be noted that this technique can also be used as accurate thickness measurement. This was found by measuring the layer thickness after selectively etching. The thickness measurements agree with a maximum deviation of ±0.3% in the simulated layer thickness.

In conclusion, we have calculated an empirical composition dependent third order interpolation matrix for the dispersion of the refractive index of the (AlGaIn)As material system with excellent accuracy and an essential improvement for the design of (AlGaIn)As-based opto-electronic devices like VCSELs can be expected. A more detailed description can be found in [4].

Fig. 2: A comparison of refractive index dispersion data for various effective $X$ fractions of $(Al_{0.48}InAs)_X(Ga_{0.47}InAs)_{1-X}$ is illustrated. The dashed-dotted line shows the fit after [2], the dashed line corresponds to [3], whereas the straight line represents our data.

High-efficiency devices and room temperature continuous wave operation of injectorless quantum cascade lasers using multi-alloy structures

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With the previous development of low threshold injectorless quantum cascade lasers using multi-alloy structures [1], the threshold power densities have strongly decreased from 70 MW/cm\textsuperscript{3} down to 38 MW/cm\textsuperscript{3} [1][2], the lowest value reported so far. In combination with an improved thermal performance, indicated by the characteristic temperature, the possibility for continuous wave operation seemed feasible. Another issue was the overall power efficiency, defined as the optical output power divided by the electrical input power, which only reached 2\% although the threshold performance was excellent.

Recent work on injectorless quantum cascade lasers focused on the development of a suitable continuous wave process and setup technology. To reduce the dissipated electric power, a 4-13 \textmu m narrow ridge waveguide process was developed. The narrow ridges were formed by a combination of dry etching and wet etching for smoothening the sidewalls. Because of the long emission wavelength, the passivation had to be created from 400 nm of silicon nitride, having lower losses, and a 350 nm silicon dioxide layer, with high losses, but stronger attenuation of the mode towards the metallization (figure 1 for process). Thermal simulations indicate that this process yields roughly 7 K/W and for reaching continuous wave operation at room temperature, a thermal resistance less than 10 K/W is required. With the standard setup already adding 15 K/W to the device, a more professional setup technology had to be developed, using electroplated copper heat sinks, with prede-

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\textbf{Fig. 1:} 4 \textmu m narrow ridge waveguide process (a) and continuous wave performance of 4.1 mm long devices with 4 and 9 \textmu m wide mesas (b). The small inset shows the threshold performance for pulsed (blue), continuous wave with standard setup (orange) and advanced setup (red)
posited thin eutectic AuSn layers for annealing. The previously developed design [1] reached cw operation temperatures above room temperature [3], with a maximum efficiency of 0.7 % at 280 K and a threshold current density of 1.75 kA/cm² at 293 K. Figure 1b summarizes the results for a broad device (9 µm) and a narrow one (4 µm), with overall thermal resistances of 10 and 16 K/W respectively.

Besides continuous wave operation, the performance of injectorless quantum cascade lasers was investigated. For improving the slope efficiency, the upper laser state lifetime was increased by implementing more InAs spikes. While three or more spikes failed due to strain relaxation, the implementation of two spikes (see figure 2) within the optical transition well improved the output power to values to 1.2 W per facet. The overall efficiency, corrected for the collection efficiency of the optical measurement system, increased to 7.1 % at 300 K, while efficiencies at 80 K are estimated to be above 30 %. Figure 3 summarizes the high performance device.

In conclusion, we have demonstrated the performance improvement by multiple InAs peaks and the continuous wave capabilities of injectorless quantum cascade lasers. Further development steps could be overgrowth for better thermal performance, distributed feedback resonators for gas sensing applications and long wavelength high performance devices in the transmission window between 8 and 12 µm.

Fig. 2: Active region using two InAs spikes within the transition well, designed for an emission wavelength of 5.4 µm

Fig. 2: Output power (red), overall efficiency (green) and inset with threshold performance

Terahertz laser sources by difference-frequency generation in Mid-IR quantum cascade lasers

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Since the first demonstration of quantum cascade laser (QCL) in 1994, tremendous improvements in the field have been achieved. The mid-IR (4 µm – 15 µm) spectral range has been covered by room-temperature operating QCLs, whereas the terahertz (THz) spectral range still lacks compact, electrically pumped, laser sources. Despite recent progress in the field, existing devices still require cryogenic cooling, which makes device applications limited. An alternative way to produce room-temperature radiation is to use difference-frequency generation in a nonlinear crystal pumped by two mid-IR beams. Such conventional nonlinear optics approach would require bulky crystals and complex setups, unsuitable for most applications. However, recently it has been shown that intersubband transitions in coupled quantum wells posses giant nonlinear susceptibilities and a room-temperature operating device based on intracavity difference-frequency generation in mid-IR QCLs has been demonstrated [1]. The best device reported so far had peak power of 7 µW at 80 K and 300 nW at RT with a conversion efficiency of 5 µW/W², which needs to be improved in order to be useful for applications.

Here we propose a novel device concept, which would allow boosting the nonlinear susceptibility and therefore the conversion efficiency up to mW/W² level [2]. The main idea is to use a separately designed nonlinear section, grown on top of a two-color mid-IR QCL active region. The band structure of such nonlinear section is sketched in Fig. 1 (a).

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nonlinear susceptibility value can reach up to $10^6 \text{ pm/V}$ offering conversion efficiencies in the mW/W$^2$ range. The schematic device cross-section is shown in Fig. 1 (c).

Our device structures were fabricated by first growing the two mid-IR QCL active regions and the nonlinear section by solid-source MBE, followed by structuring of the nonlinear section by selective wet-chemical etching. Afterwards, the devices were transferred into MOVPE reactor, where the InP waveguide and top GaInAs contact layers have been overgrown. The wafers were then processed by deep etching the ridges via Cl based reactive ion etching, followed by combined Si$_3$N$_4$/SiO$_2$ insulation deposition by PECVD and sputtering. After opening the contact windows with CF$_4$ plasma etching, top metal contact of Ti/Pt/Au was applied. Finally the devices were thinned down to approximately 120 µm and Ge/Au/Ni/Au back contact was evaporated.

**Fig. 2:** Recorded emission spectra of THz signal (left) and corresponding mid-IR pump signals at three different driving currents at 78 K.

Cleaved and mounted devices were tested under pulsed operating conditions. Experimental emission spectra of THz and mid-IR pumps are shown in Fig. 2 (a) and (b) respectively. First generation devices operated up to room temperature, exhibiting peak power of 105 µW at 80 K with a conversion efficiency of around 30 µW/W$^2$ [3]. The first results are encouraging, and further improvements, such as waveguide optimization for phase-matching and better out coupling are on the way.

In conclusion we have demonstrated a novel concept for a compact, room-temperature THz laser source using intracavity difference-frequency generation in a separated nonlinear section by mixing two mid-IR beams from the dual-color QCL active region.


Financial support by the Excellence Cluster “Nanosystems Initiative Munich (NIM)” is gratefully acknowledged.
How periodic are terahertz quantum cascade lasers?

Tillmann Kubis and Peter Vogl¹

Quantum cascade lasers (QCLs) are unipolar lasers where the upper and lower laser level lie within the conduction band of specially designed semiconductor multi quantum wells. Electrons are made to enter such a quantum well region by an externally applied bias. Initially, they occupy the upper laser level and emit a photon because a barrier prevents them from leaving the region by tunneling or hopping. Once they reach the lower level, they can quickly leave the quantum well region by emitting a phonon and traversing to the adjacent quantum well region or, finally, a contact. In order to enhance the efficiency of such a QCL, it consists of hundreds of such quantum well regions that are designed in such a way that electrons that leave one period end up in the upper laser level of the adjacent one. Thus, ideally the distribution of electrons is exactly the same in each period. In practice, however, one finds that only a few percent of the QCL periods actually contribute a photon which indicates that the electrons are not distributed among the quantum well periods in a strictly periodic fashion.

We have developed a self-consistent non-equilibrium Green's function method (NEGF) for stationary charge transport in open THz-QCL devices. Our method does not impose field-periodic boundary conditions. Instead, the electrons can enter and leave the device via travelling eigenstates of the semi-infinite leads. This allows us to capture non-periodic phenomena as well as investigate spatially extended, energetically high lying states. Details of the method are given in [1].

In this work, we are able to illustrate the formation of electron distributions in resonant phonon THz-QCLs that are not commensurable with the QCL device periods and theoretically study the impact of such distributions on the device current density and the optical gain.

![Graph](image)

**Fig. 1:** Calculated current density as a function of applied bias voltage per period of the QCL of [2] at 40K and a sheet doping density of $1.9 \times 10^{10}$ cm$^{-2}$. and comparison to experiment [2]. The results of the single-period model (line) and two-periods model (squares) agree well with experiment (dots).

We consider electron transport in the resonant phonon THz-QCL structure of [2] which consists of 271 identical periods of GaAs and Al$_{0.15}$Ga$_{0.85}$As layers of the widths (30) 92 (55) 80 (27) 66 (41) 155 Å. The values in parentheses indicate the Al$_{0.15}$Ga$_{0.85}$As barriers and the underlined well is the only doped region with a doping density of $1.9 \times 10^{10}$ cm$^{-2}$. We calculate charge transport in terms of two models. In one model, we treat a single device period as active device, whereas the other model includes explicitly two device periods. In Figure 1, we depict the calculated (black) and experimental (gray dots) I-V char-

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acteristics of this QCL. Both models agree with one another and reproduce the experimental data of the QCL up to the threshold bias.

In figure 2 (a), we show the contour plot of the spectral function at vanishing in-plane momentum in two adjacent periods of the QCL for a bias of 52 mV/period. The maxima of this spectral function correspond to resonant electronic states. The states associated with the first and second QCL period are labelled by numbers and primed numbers, respectively. The carrier distribution markedly deviates from the geometric QCL periodicity. This can be seen most clearly in Figure 2 (b) by a contour graph of the local energy resolved current density $j(z,E)$. The function $j(z,E)$ shows spatially constant (i.e. horizontal) stripes in regions where the electrons propagate without dissipating energy. Disruptions of these horizontal stripes mark positions where LO-phonons get emitted. The figure shows that the number of emitted LO phonons is no longer equal for adjacent QCL periods. When the electrons have passed the first period and traversed a potential drop of 52 meV, they have emitted only one LO-phonon of energy 36 meV. After electrons have traversed the second period, they have emitted a total of 3 LO phonons across 2 QCL periods and are finally fully thermalized. This process is repeated in the subsequent QCL periods such that we obtain a commensurable charge distribution with period two. A consequence of this incomplete carrier thermalization is a significant reduction in the occupation inversion and the optical gain in every other period.

Fig. 2: (a) Calculated conduction band profile (line) and contour plot of the energy and spatially resolved spectral function $A(z,E)$ at vanishing in-plane momentum in two adjacent periods of the QCL at a bias voltage of 52 mV per period in the relevant energy range. The dotted line marks the boundary between first and second period. (b) Calculated spatially and energy resolved current density $j(z,E)$ of the QCL in (a).


Supported by the DFG, SFB 631, Austrian SFB IRON, Munich Nano Initiative
Biophysics and Bioelectronics
Diamond-based biosensor arrays

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Most biosensors are hybrid devices composed of a bioreceptor, responsible for the selective recognition of the analyte, and a transducer, transforming the signal generated by the recognition reaction into one suitable for further processing and electronic readout. The basic concept of an amperometric biosensor is shown schematically in Fig. 1. In this type of biosensor an enzyme serves as bioreceptor. Enzymes are proteins catalyzing chemical reactions – usually very selectively. For an enzymatic redox reaction, the associated charge transfer can be detected by measuring a current via the diamond electrode. The nanocrystalline diamond (NCD) electrode serves as the transducer. Diamond exhibits outstanding properties which render it very well suited as an electrode material interfacing biological systems [1]. Mechanical stability, chemical inertness, and the wide electrochemical window allow application in harsh environments.

At the same time diamond shows excellent bio-compatibility. Nanocrystalline diamond films can be grown on various substrates by microwave plasma-assisted chemical vapor deposition (MPCVD). The electronic properties can be defined by doping with boron or phosphorous, which is achieved by mixing suitable precursor gases into the plasma during the growth. By selective nucleation with diamond nanoparticles, patterned diamond structures can be grown. Fig. 2 shows a scanning electron microscope (SEM) image of a quadratic thin film (thickness 300 nm) of highly boron doped NCD on a platinum contact. The area covered by the boron doped, and hence conductive, NCD represents the active area of the sensor, on which in a later step the enzymes will be immobilized. The rest of the surface will be covered by a layer of intrinsic, and thus insulating, NCD. Many of these small electrodes, which can all serve as individual biosensors, are integrated on a single sensor chip.

Controlled immobilization of enzymes is essential in the design of this type of biosensors. The most common grafting technique is covalent attachment using self assembled monolayers (SAMs) of linker molecules. In cooperation with the Wacker Chair of Macromolecular Chemistry we investigate protein immobilization via polymer brushes. In a first

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step, polymers are grown on the diamond surface by self-initiated photo-grafting and photo-polymerization. With this technique no extensive pretreatment of the surface is necessary to initiate polymerization – a termination with hydroxyl groups is sufficient. Proteins can be directly tethered along polymers built up from monomers exhibiting suitable functional groups. Alternatively, in a second step side chains can be generated by graft-polymerization, providing additional functionalities [2], for example conductive groups to enhance charge transfer from proteins immobilized along the polymer far away (e.g. 10 nm) from the NCD electrode. The advantages of this approach include providing a more natural environment for the proteins, a higher loading density of proteins per surface area (hence a possibly stronger signal), and a more tunable system with respect to properties such as hydrophilicity, grafting density, additional functionalities, etc. It is crucial for this approach that proteins, in particular enzymes, can be immobilized along the whole length of the polymers. Hence, large molecules must be able to penetrate into the layer formed by the polymer brushes. The inset in Fig. 3 shows a fluorescence microscope image of a NCD surface to which green fluorescent proteins (GFP) were coupled via the free carboxylic acid sites of a poly(methacrylic acid) (PMA) polymer brush patterned onto the surface. The brush exhibits a grafting density gradient created by e-beam carbon deposition (EBCD) templating. As Fig. 3 illustrates, the height profile of the polymer brush follows the grafting density defined by the EBCD dose, even though the contour length of the individual polymers is similar. The fluorescence intensity also follows the height profile measured by an atomic force microscope (AFM), as shown in Fig. 4, which confirms that GFP is incorporated into the whole polymer brush. In conclusion, we demonstrate the controlled immobilization of proteins, using the example of GFP, into polymer brushes. This paves the way for further investigation of the system (protein / polymer brush / NCD) with respect to charge transfer characteristics, enzyme activity, and long term stability.

Fig. 3: PMA polymer brushes with coupled GFP on NCD with EBCD templates. Left/right: low/high EBCD dose, low/high grafting density, profile height smaller/equal contour length, less/more GFP per surface area
Inset: Fluorescence image of the gradient

Fig. 4: Height and fluorescence profiles along the GFP-modified polymer brush gradient in the Fig. 3 inset

Organic thin-film transistors for sensing in electrolytic environments

Felix Buth, Deepu Kumar, Martin Stutzmann, and José A. Garrido

Organic semiconductors are low-cost alternatives to their inorganic counterparts, making it economically interesting to use them in disposable devices, such as sensors. Especially interesting for medical and biological applications are thin-film transistors, operating in an electrolytic environment. First examples of such devices have been presented in the 1980s with conducting polymers as the active layer [1]. These solution-gated transistors work, however, via a chemical doping process. The discovery of water stability of some semiconducting small molecules [2] has initiated research on thin-film sensors using field-effect transistor with a conventional inorganic back gate. In our work we investigate the use of small molecules in solution-gated transistors.

As water stable molecules we used sexithiophene (6T). Upon vacuum evaporation on inert substrates hydrophobic polycrystalline thin films are formed (Fig. 1). The films show a typical mixture of islands consisting of either upright standing or lying down molecules (bright regions in Fig. 1). The hydrophobicity of the film is expected to prevent substantial penetration of water into the organic layer and thereby reduce degradation. Furthermore, it facilitates the formation of an electrical double layer at the semiconductor electrolyte interface. Using cyclic voltammetry (CV) and electrochemical impedance spectroscopy we showed that the capacitance of this double layer is of the order of 2 µF/cm² (Fig. 2), which agrees well with literature values for double layers formed between hydrophobic semiconductors and aqueous electrolytes. This high interfacial capacitance enables low-voltage operation of the transistors, a requirement for electrolyte operation. At this point it has to be mentioned that this analysis is only valid if our device works via an electrical field effect, which agrees well with all our data, but we can so far not completely rule out a chemical doping effect. However, the absence of peaks in the CV between -0.7 and 0.2 V excludes the doping via redox-reactions in this voltage regime.

Using Ti/Au bottom-contacts on an 50nm thick 6T film. The inset shows a contact angle measurement of this film revealing its hydrophobicity.

Fig. 1: Cyclic voltammogram of a sexithiophene (chemical structure shown) thin film on gold in a KCl electrolyte.

Fig. 2: Atomic force micrograph of a 6T film. The inset shows a contact angle measurement of this film revealing its hydrophobicity.

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inert oxidized diamond surface we have fabricated thin-film transistors with a channel length and width of 20µm and 200µm, respectively (Fig. 3a and b). The application of an additional SU8 chemical resist layer on top of the contacts reduces the leakage currents to the gate electrode. The transistor curves measured with this geometry resulted in field-effect mobilities on the order of 1x10^{-2} cm²/Vs (Fig. 3c). This rather low mobility can be explained by the rough interface (see Fig. 1) between the electrolyte and the organic semiconductor at which the conductive channel is formed. In addition, the creation of defects by interpenetration of water or ions into the channel needs to be taken into account. We already observed a smoother morphology for films grown at slightly elevated substrate temperatures. These films should consequently yield higher carrier mobilities and will therefore be the topic of future research. In terms of sensitivity the direct contact of the channel with the electrolyte is, however, expected to be an advantage over the back gate architecture. Furthermore, the electrolyte gated devices have a simpler design since no additional gate dielectric is needed.

![Fig. 3: a) Schematic layout of the device. b) Optical image of an array consisting of eight transistors. c) Characteristics of a 6T transistor operated in KCl electrolyte, showing a clear gate effect and current saturation at higher voltages applied between source and drain contacts.](image)

In conclusion we showed that small molecular organic semiconductors can be used in electrolyte gated thin-film transistors, with the goal of using them in sensing devices. Further work is, however, needed in order to completely understand the operating mechanism of these devices.


Low frequency noise in diamond solution-gate field effect transistors

Moritz Hauf, Lucas Hess, John Howgate, Markus Dankerl, Martin Stutzmann, and José A. Garrido

Hydrogen-terminated diamond surfaces have been extensively studied in the past few years as transducer materials in biosensing applications. The surface conductivity of hydrogen-terminated diamond can be modulated in an electrolyte by a voltage applied between a reference electrode and the diamond, allowing for the fabrication of diamond solution-gate field effect transistors (SGFETs). In the past, these have been shown to be suitable sensors for changes in pH and ionic strength of the electrolyte as well as for the extracellular detection of action potentials from cells [1].

In the challenge of creating not only highly specific but also extremely sensitive devices, a good understanding of the electronic noise is of utmost importance. Unlike silicon devices, the diamond SGFETs do not require a gate oxide, which is assumed to introduce low frequency noise in silicon based devices due to trapping and detrapping of charge carriers. Here, we investigate the origin of the low frequency noise in diamond SGFETs.

Device structures are fabricated on hydrogen-terminated single-crystalline diamonds by means of conventional photolithography. Contacts are sealed against the electrolyte by photoresist, leading to access regions between the gate channel and the gold contacts.

Fig. 1: a) Transistor response to change in $U_{GS}$ (red line). The deviation from the ideal behavior (blue line) is attributed to the presence of access regions. b) Noise power spectral density of a diamond SGFET (red line) revealing a $1/f$–type noise. c) $U_{GS}$ dependence of the power spectral density at 1Hz

Fig. 1a) shows the dependence of the drain-source current $I_{DS}$ on the gate voltage $U_{GS}$. A saturation is observed for highly conductive channels as the access regions (hydrogen-terminated diamond covered with photoresist) start to dominate the overall conductivity. The current noise spectra reveal a clear $1/f$ dependence in the low frequency regime (Figure 1 b) which can be evaluated according to Hooge’s empirical formula:

$$S_{i} = \alpha_{H} \frac{I^{2}}{N f}$$

Here, $S_{i}$ is the power spectral density, $\alpha_{H}$ the Hooge parameter, $N$ the number of charge carriers, $I$ the current and $f$ the frequency. Figure 1 c) shows the dependence of $\alpha_{H}/N$ on the

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gate voltage, revealing a decrease with $U_G^{-3}$ and a subsequent saturation for high gate voltages. This dependence can be explained, following the argumentation of Peransin et al. [2], when the contribution from the access regions is taken into account. The overall noise is considered to be the sum of an independent noise contribution from the access region (not gate voltage dependent) and the diamond channel (gate voltage dependent). In a similar way the total resistance is considered being the sum of the access resistance and the channel resistance. This way, we are able to extract the real Hooge parameter of the hydrogen-terminated diamond channel.

Using the van-der-Pauw structure, the dependence of the number of charge carriers $N$ on the gate voltage is determined by in-electrolyte Hall measurements showing a linear dependence (Figure 2 a).

![Image](image_url)

**Fig. 2:** a) charge carrier concentration $n_h$ dependence on the gate voltage. b) Hooge parameter $\alpha_H$ dependence on $n_h$. For reference the slope of $n^{-1}$ is shown in red. c) Comparison of the effective gate noise for different SGFETS.

Figure 2 b) reveals an inverse dependence of the Hooge parameter $\alpha_H$ on the carrier concentration $n_h$, which we attribute to number fluctuations of the charge carriers [3]. Unlike in silicon FETs the trapping/detrapping of carriers at the interface of the gate oxide that leads to the number fluctuations cannot occur in diamond SGFETs. Therefore, we suggest that defects related to dislocations in the diamond crystal are responsible for trapping/detrapping of carriers.

For sensing applications the effective gate voltage noise is of significant interest as it determines the resolution limit of the device. The effective gate voltage noise $S_{UG}$ is derived from the current noise $S_I$ by taking into account the transconductance $g_m$: $S_{UG} = S_I g_m^2$.

Figure 2 c) shows that diamond SGFETs perform considerably well in comparison to other commonly used materials like silicon and GaN. In the sensing relevant frequency range, from 1 Hz to 10 kHz, the gate voltage fluctuations of diamond SGFETs can be calculated to 16µV. The rapid improvement of the quality of CVD-grown diamond, especially regarding a decrease of the dislocation density, is expected to further decrease the low frequency noise of diamond SGFETs.


Supported by the Graduate School Materials Science of Complex Interfaces, CompInt
Graphene solution gate field effect transistors

Markus Dankerl¹, Moritz V. Hauf, Andreas Lippert, Lucas Hess, Martin Stutzmann, and Jose A. Garrido

Graphene is a recent contender for the material of choice in the field of biosensors and bioelectronics. While silicon is the most widely used material in this field due to its technological maturity, it has serious drawbacks regarding e.g. stability for operation in the relevant physiological conditions. Graphene, like other carbon materials such as diamond, exhibits superior chemical and electrochemical stability, combined with its unsurpassed electrical performance.

We have fabricated graphene solution gate field effect transistors (SGFETs). Here, the monolayer graphene sheet (see Fig. 1a) is gated with an electrolyte solution and can be operated both in the hole and the electron accumulation region.[1] The gate voltage $U_G$ is applied between a reference electrode in the electrolyte and a contact in the graphene, thereby fixing the potential drop between the graphene Fermi level and the reference electrode. Shifting the Fermi level with the gate potential below or above the Dirac point accumulates holes or electrons in the graphene sheet, respectively (see Fig. 1b).

For these devices we have chosen epitaxial graphene on a SiC substrate. This material system is scalable and can therefore take full advantage of our photolithographic processing method. The graphene areas were defined by etching away undesired areas with an oxygen plasma. The bond pads were Ti/Au stacks directly on the SiC substrate. An additional Au layer established the contact by overlapping onto the graphene. In order to prevent contact between the metal contacts and the electrolyte, the contacts were covered by a chemically stable photoresist, SU-8, leaving only the graphene exposed to the electrolyte. This processing method allows the fabrication of micrometer range structures on a large scale (see Fig. 1c).

Figure 1: a) In-situ STM image revealing the morphology of the as-grown epitaxial graphene, with large terraces of single layer graphene (ML) and patches of bilayer islands (BL). b) Schematic drawing representing the modulation of the carrier density in the graphene film. c) Optical micrograph of graphene SGFETs. The Ti/Au structures (yellow) represent drain and source contacts for the transistors and contacts for the van der Pauw structure in the centre.

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Electrical characterization of these devices in electrolyte (10 mM PBS adjusted to 50mM ionic strength with KCl) shows the expected behavior according to the operation principle described above (see Fig. 2a). Both n- and p-type conductivity is observed dependent on the gate voltage with a V-shaped minimum around the Dirac point. In light of the processing technology described above, the region, termed access region, of the graphene sheet directly in the vicinity of the contacts and beneath the passivating photoresist is of significant importance. This area of the graphene is obviously not in contact with the electrolyte, therefore not gated, and in a low conducting state. Therefore it contributes an access resistance to the overall resistance of the graphene channel. In order to quantify this effect we have measured the conductivity employing the van-der-Pauw structure (see Fig 1b), thereby eliminating contact resistances with this four point probe method. This sheet conductivity together with a serial access resistance gives the calculated resistance represented by the solid lines shown in Figure 2a and lets us determine the access resistance. The effect of this access region can also be seen in the transconductance $g_m$ (see Fig. 2b). As a high transconductance, meaning a high modulation of the drain source current $I_{DS}$ with the gate potential, is obviously desirable for a sensor, Figure 2b shows the need to minimize the access resistance in order to gain sensitivity.

**Figure 2:** A) Drain-source current curves (for $U_{DS} = 100$ mV) measured (solid symbols) for two graphene transistors at different gate voltages. The solid line represents the estimation of the drain-source current from the 4-point resistivity measurements and the contribution of the access resistance. B) Effect of the access resistance on the transconductance. The solid symbols correspond to the so-called extrinsic transconductance, i.e. $\partial I_{DS}/\partial U_G$, which represents the modulation capability of the gate. Due to the effect of the access resistance, the extrinsic transconductance is strongly reduced with respect to the intrinsic transconductance given by the open symbols.


Supported by EU-FP6 : Diamond to Retina Artificial Microinterface Structures (DREAMS)
Graphene biosensors: Modeling the hydrophobic graphene/electrolyte interface

Stefan Birner¹, Andreas Lippert¹, Markus Dankerl¹, José A. Garrido, Nadine Schwierz², Roland R. Netz², and Peter Vogl¹

Graphene, with its unique combination of physical and electronic properties, is a very promising material not only for electronics, but also for biosensor and bioelectronic applications. In this respect, a quantitative understanding of the sensitivity of graphene solution-gated field effect transistors (SGFET) capable of operation in aqueous environments is important. Here, we present results of numerical calculations on the variation of the charge carrier concentrations under electrolyte gate control. The effect of the solution-gate potential on the electronic properties of graphene is explained using a model which considers the microscopic structure of water at the graphene/electrolyte interface.

We calculate self-consistently the spatial charge and electrostatic potential distribution in the graphene/electrolyte system by solving the nonlinear Poisson equation. The model has been implemented into the nextnano software [1]. It includes a spatially varying static dielectric constant in the electrolyte which is proportional to the water density variation according to Ref. [2], and varies from $\varepsilon_r = 1$ at the interface to $\varepsilon_r = 78$ further away from the interface. The distributions of the different ions (Na⁺, Cl⁻) in the electrolyte are calculated using an extended Poisson-Boltzmann approach that takes into account recently published [2] ionic potentials of mean force (PMFs). The fitting functions of the PMFs correspond to hydrophobic (i.e. nonpolar) solid/liquid interfaces based on data obtained from atomistic molecular dynamics simulations [2]. The PMFs have the effect of repelling the ions from the interface and are zero at distances larger than 1.4 nm from the interface.

We compare this approach with the traditional Poisson-Boltzmann equation where no PMFs are employed while assuming a constant value of $\varepsilon_r = 78$ for the static dielectric constant of the electrolyte. We find significant differences for the spatial distribution of the resulting ion density (Fig. 1), as well as for the potential distribution, sheet carrier density and capacitance (Fig. 2). The variation of the electrostatic potential across the graphene/electrolyte interface is shown in Fig. 2a) for two different values of the applied gate potential $U_G$ revealing a significant potential drop in the electrolyte for the extended Poisson-Boltzmann model, whereas in the

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In the case of the simple Poisson-Boltzmann (PB) model, the potential drop in the electrolyte is almost negligible and most of the potential drops in the graphene layer, i.e. the electrostatic potential in graphene approximately equals $U_G$. This situation resembles the “quantum limit” condition where a shift in $U_G$ corresponds directly to a shift of the Fermi level with respect to the Dirac point, i.e. $E_{F, \text{graphene}} = eU_G$. In the case of the extended PB model, an important potential drop occurs in the electrolyte, which strongly reduces the electrostatic potential in the graphene film. The carrier density versus applied gate voltage obtained for the two different models, together with the quantum limit, is shown in Fig. 2b. Here, the hydrophobic model (extended PB) leads to results very close to the measured electron and hole densities in graphene [3]. The results of the total capacitance calculations ($C = \partial Q / \partial U_G$) using the standard PB model and the extended PB model are compared in Fig. 2c) to the case of the quantum limit. Here, the grey line corresponds to a simple plate capacitor model (width $d = 0.32$ nm, $\varepsilon_r = 1$). Consequently, at large voltages the plate capacitor model describes nicely the hydrophobic double layer capacitance whereas for small voltages the quantum capacitance of graphene dominates. The interfacial capacitance is a series capacitance of the quantum capacitance of graphene and the double layer capacitance of the electrolyte. Thus, the double layer capacitance at the graphene/electrolyte interface is about 3 µF/cm². In conclusion, a proper description of the graphene/electrolyte interface, which considers the effect of the solid surface on the dielectric properties of the water, can explain the experimentally measured low value of the double layer capacitance.

Functionalization of silicon carbide with structured polymer brushes

Marin Steenackers¹,², Naima A. Hutter², Rainer Jordan³, Martin Stutzmann¹, and Ian D Sharp¹

Silicon carbide (SiC) is a very promising candidate for biotechnology and biosensing applications. This wide band gap semiconducting material is mechanically extremely hard, optically transparent, chemically inert, and biocompatible. In the perspective of using SiC for biomedical applications, efforts are made for the direct immobilization of biomolecules on SiC electrodes. While most studies on the biofunctionalization of SiC are based on monolayer approaches, it has been shown that thin biocompatible polymer brushes can behave as ideal soft interfaces between biochemical systems and rigid substrates such as semiconductors. Large biomolecules can penetrate into the soft polymer brush layer and interact with binding partners in a three-dimensional arrangement. This system thus resembles real biological environments much more than the case of direct immobilization on flat and rigid surfaces (potentially causing denaturation of proteins). Furthermore, such polymers allow the design of biosensors with higher loading capacities and enhanced sensitivity compared to conventional self-assembled monolayers.

Here, we describe a straightforward method for the preparation of homogeneous as well as (nano)-structured polymer brushes on both hexagonal and cubic silicon carbide substrates. By atomic force microscopy (AFM) and Fourier transform infrared spectroscopy (FTIR), we have shown that the self-initiated photografting and photopolymerization (SIPGP) of various vinyl monomers (styrene and (meth)-acrylate derivatives) allows the preparation of polymer brush layers directly and selectively onto C-OH terminated 6H-SiC surfaces as well as (111) 3C-SiC surfaces in a single step reaction [1,2]. In contrast, the analogous reaction on Si-OH terminated (0001) 6H-SiC as well as on Si-OH terminated (100) 3C-SiC does not occur (Fig. 1). This reactivity contrast for SIPGP is explained by density functional theory calculations which showed a higher bond dissociation energy for hydrogen abstraction at Si-OH relative to C-OH terminal sites.

Structured polymer brushes with defined features on the sub-micrometer scale were prepared by the carbon templating (CT) technique [3] on Si-OH terminated (0001) 6H-SiC and (100) 3C-SiC surfaces (Fig. 2). The CT technique allows the synthesis of well-defined micro- and nanostructured polymer grafts with near-molecular precision. A stable carbon template layer is locally formed by electron beam carbon deposition (EBCD) and amplified by the formation of a grafted polymeric layer by means of SIPGP of vinyl monomers. In contrast to all existing methods, the carbon-

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templating approach is a SAM-, initiator- as well as photoresist-free process which avoids complex and tedious multi-step reactions and minimizes surface contamination. One major advantage of this approach is that the resulting polymer grafts are thermally and chemically stable (instability is a major drawback for commonly used monolayer-based systems) and allow us to introduce complex organic and biomolecules within the polymer layer by polymer analogue reactions. The preparation of structured functional polymer brushes can be used for the patterned attachment of, for example, proteins or other biomolecules to create sensor arrays directly onto biocompatible SiC surfaces. Furthermore, the CT-technique allows control of not only the 2D locus of the polymer brushes and thus the grafted biomolecules, but also, by varying the locally applied electron dose, control of the grafting density of functional groups.

Currently, work is in progress to investigate further modifications of structured polymer brushes on SiC with biomolecules. Furthermore, we are studying - in collaboration with the research group of Kian Ping Loh at the National University of Singapore - the functionalization of epitaxial- as well as CVD graphene by a similar approach. With its unique mechanical, electrical and optical properties, graphene has potential applications in various fields such as electronics, single molecule gas detection, (bio)sensors, and microelectromechanical systems.

Fig. 2: AFM image of an array of structured polymer brushes created by the SIPGP of styrene (polymerization time 16h) on different carbon template structures on a Si-OH terminated 6H-SiC (0001) surface.

Self-assembled alkene-derived monolayers on silicon carbide

Marco Hoeb¹, Sebastian Schoell, Marianne Auernhammer, Martin Stutzmann, Martin S. Brandt, and Ian D. Sharp

Chemical surface modification of silicon carbide (SiC) has received increasing interest in recent years because of its unique chemical and structural properties of this material. SiC is a technologically important semiconductor with bandgap energies ranging from 2.4 eV for 3C-SiC to 3.3 eV for 4H-SiC, thus offering the possibility for bandgap engineering and tuning of charge transfer between SiC and organic and inorganic epilayers.

Besides silicon, silicon carbide is the only compound semiconductor that forms a native silicon oxide overlayer. For electron transport studies across the interface, this surface oxide layer is an impediment and can be stripped off via etching in hydrofluoric acid (HF). For silicon, the removal of the oxide results in a perfectly hydrogen-terminated interface [1]. This hydrogenated surface is reactive towards alkenes – a process called hydrosilylation first described by Linford [2]. The reaction mechanism is based on either thermal or light induced cleavage of the surface hydrogen that induces a carbon radical in the alkene chain. This is the start of a chain-reaction that forms a dense Si-C bound organic monolayer of alkene molecules.

Here we describe the transfer of this process to 6H-SiC. For silicon carbide, there is no known wet-chemical process to obtain hydrogen saturation of the surface, and etching in HF yields mainly OH-terminated surfaces [3,4]. Nevertheless, recent reports have suggested that alkene-derived self-assembled monolayers also form on hydroxyl-terminated, oxidized silicon surfaces [4]. This observation applies also to alkene-derived monolayers on both (0001) Si- and (0001) C-polar 6H-SiC surfaces. Subtle, but potentially important differences exist between the two crystal faces.

Functionalization of the HF-etched SiC samples was achieved using 1-octadecene. The alkene was dried under reflux, vacuum distilled and stored under argon. The dried reaction vessel was typically filled with 0.5 ml of neat octadecene. For thermal alkylation the vessel was immersed into an oil bath at 200°C for two hours, for UV-functionalization the samples were illuminated for typically 45 min with a mercury arc lamp. The samples were finally sonicated with several solvents and blown dry under nitrogen. The following results refer to thermally functionalized samples, but are similar on UV-functionalized samples.

Figure 1: (a) Thermal and (b) UV-light induced functionalization of SiC surfaces. In both cases the reaction starts from a mainly hydroxyl-terminated, polar surface that forms covalent bonds with the liquid alkene.

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X-ray photoelectron spectra (XPS) of the C(1s) and O(1s) core levels obtained immediately following HF etching of both surfaces are presented in Fig. 2(a & b). Both etched surfaces exhibit strong substrate peaks. Inspection of the O(1s) core level spectra of both etched substrates, shown in Fig. 2(c & d), confirms the presence of oxygen on the surfaces. Following thermal functionalization, an additional C(1s) peak appears with a positive chemical shift of 2.0 eV (1.8 eV) on the Si-face (C-face), which is characteristic of adsorbed alkyl chains. The substrate signal intensity simultaneously decreases due to attenuation by the shielding organic layer. Accordingly, the oxygen peaks is attenuated by the organic monolayer by roughly the same factor. This result provides evidence that oxygen participates directly in molecular bonding and a Si-O-C (C-O-C) bond between substrate and molecule is formed.

In conclusion, we have demonstrated wet-chemical alkoxylation of HF-treated 6H-SiC surfaces. Photoelectron spectra confirm that, similarly to oxidized silicon, alkenes can graft via either thermal or UV activation to SiC substrates with interfacial oxygen bridging configurations.

![Fig. 2: XPS spectra of alkoxylated Si- and C-face SiC before and after monolayer formation. (a) Si-face C 1s spectra, (b) C-face C 1s spectra, (c) Si-face O 1s spectra, and (d) C-face O 1s spectra. All spectra are normalized to maximum intensity of the C 1s signal of the HF-etched C-face sample.](image)

Fabrication of metalized nanopores in silicon nitride membranes for single molecule sensing

Ruoshan Wei, Daniel Pedone, Andreas Zürner, Markus Döblinger, and Ulrich Rant

Single nanometer-sized holes in otherwise impervious membranes constitute simple devices to analyze single molecules. Artificially fabricated (solid-state) nanopores allow to engineer the size, shape and chemical properties of the pore, and thus to equip the pore with desired functionality.

![Fig. 3: A) TEM images of nanopores in SiN membranes. B) The pore diameter $d_{\text{SiN}}$ can be tuned by the e-beam exposure dose. C) TEM tomography reconstructions of a SiN nanopore.](image)

**Fig. 3:** A) TEM images of nanopores in SiN membranes. B) The pore diameter $d_{\text{SiN}}$ can be tuned by the e-beam exposure dose. C) TEM tomography reconstructions of a SiN nanopore.

Here we demonstrate the fabrication of metalized nanopores in SiN membranes with widely available semiconductor process technology. We employed electron beam lithography and reactive ion etching to fabricate nanopores with diameters down to 10 nm in a SiN membrane. Figure 1A shows exemplary TEM images of pores of various diameters. The resulting pore diameter $d_{\text{SiN}}$ can be controlled with very good reproducibility by adjusting the e-beam exposure dose $D$ (Fig. 1B). Furthermore, we investigated the nanopore shape by electron tomography and the computed 3D reconstructions are displayed in Figure 1C. The 3D images clearly reveal a conical pore shape with a sidewall angle of 70°.

With the aim of preparing metallic nanopore structures, we evaporated thin films of Ti and Au onto pre-fabricated SiN nanopores. Figure 2A shows TEM images of the same nanopores as in Fig. 1A after the deposition of 3 nm Ti adhesion layers and 30 nm Au layers. As a result of the metal deposition, the pores shrink due to coating of the inner pore walls with metal film. In order to quantify the reduction in pore diameter in dependence of initial pore size and evaporated metal film thickness, we metalized SiN pores of different initial diameters $d_{\text{SiN}}$ with metal films ranging in thickness from 13 to

![Fig. 4: A) TEM images of nanopores after metallization with 30 nm gold. B) Average change in pore radius as a function of deposited metal thickness. C) Tomographic reconstructions of a nanopore metalized with 20 nm gold. Due to the large difference in atomic numbers between Au and Si/N, only the evaporated gold film is visible in the tomographic reconstruction (but not the SiN membrane).](image)

**Fig. 4:** A) TEM images of nanopores after metallization with 30 nm gold. B) Average change in pore radius as a function of deposited metal thickness. C) Tomographic reconstructions of a nanopore metalized with 20 nm gold. Due to the large difference in atomic numbers between Au and Si/N, only the evaporated gold film is visible in the tomographic reconstruction (but not the SiN membrane).

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45 nm. All pores were characterized by TEM before and after the metal deposition and we found that the pores shrink by a constant value \( \Delta d \) which depends solely on the evaporated film thickness \( t \). Figure 2B shows that for the metal thicknesses studied here, a linear relationship between \( t \) and \( \Delta d \) is observed. The reduction in diameter corresponds almost exactly to the geometric expectation for evaporation onto sloped sidewalls. To visualize the geometry of a metalized nanopore we used scanning transmission electron microscopy in high angle annular dark field mode (STEM-HAADF). The tomography image in Figure 2C confirms that most parts of the inner pore wall are coated with gold and the conical pore shape is preserved upon gold deposition.

The applicability of the metalized pores for stochastic sensing is demonstrated in real-time translocation experiments of single \( \lambda \)-DNA molecules. For that purpose, we passivated the pore with a self-assembled monolayer (SAM) of PEGylated alkane thiols. Figure 3A shows a representative current-time trace after addition of 2 nM \( \lambda \)-DNA. The translocation of \( \lambda \)-DNA could be observed as transient blockades of the trans-pore current.

Interestingly, the passage of DNA through PEGylated metalized pores was found to last significantly longer than through SiN or other dielectric pores, which indicates attractive interactions between the DNA and the pore walls. Despite these attractive interactions, the pore did not become irreversibly clogged by the DNA, but eventually electrophoretic forces prevailed in pulling it through the pore. Moreover, a close inspection of the current pulses reveals that the current changes are not monotonic but feature distinct jumps. We examine the current blockage ratio \( B_r \) which is defined as \( B_r = \frac{I_{\text{blocked}}}{I_{\text{open}}} \). Figure 3B shows a histogram of the blockage ratio and at least 7 separate equidistant peaks can be clearly identified. The occurrence of distinct current blockade levels and the estimation of the corresponding blocked volume suggest that the DNA enters and exits the pore in loops, which are probably extracted from a DNA “coil” outside the pore entrance (Fig. 3C).

Thus a pore concept similar to the one presented here might be suited to intentionally slow the DNA translocation, for instance, in attempts to sequence the DNA.


Supported by DFG via SFB 863, BMBF (0312031), Nanosystems Initiative Munich (NIM), and the TUM Institute for Advanced Study.
Electrically facilitated translocations of proteins through silicon nitride nanopores: Conjoint and competitive action of diffusion, electrophoresis, and electroosmosis

Matthias Firnkes, Daniel Pedone, Jelena Knezevic, Markus Döblinger, and Ulrich Rant

Solid-state nanopores bear great potential to be used to probe single proteins [1]; however, the passage of proteins through nanopores was found to be complex, and unexpected translocation behavior with respect to the passage direction, rate, and duration was observed [2]. Here we study the translocation of a model protein (avidin) through silicon nitride (SiN) nanopores focusing on the electrokinetic effects that facilitate protein transport across the pore [3]. The nanopore zeta potential $\zeta_{\text{pore}}$ and the protein zeta potential $\zeta_{\text{protein}}$ are measured independently as a function of solution pH. Our results reveal that electroosmotic (EO) transport may enhance or dominate and reverse electrophoretic (EP) transport in nanopores. The translocation behavior is rationalized by accounting for the charging states of the protein and the pore, respectively; the resulting translocation direction can be predicted according to the difference in zeta potentials, $\zeta_{\text{protein}} - \zeta_{\text{pore}}$. When electrophoresis and electroosmosis cancel each other out ($\zeta_{\text{protein}} = \zeta_{\text{pore}}$), diffusion becomes an effective and bias-independent mechanism which facilitates protein transport across the pore at a significant rate.

Fig. 1 shows the dependence of the zeta potential of avidin and the SiN nanopore surface. We measured the zeta potential of avidin in ac electric fields in solution of low salinity by phase analysis light scattering (PALS laser Doppler electrophoresis) using a Zetasizer Nano instrument (Malvern Ltd., U.K.). The nanopore zeta potential was determined by streaming potential measurements. To this end, the nanopore chip was installed in a home-built pressure cell where the chip is sandwiched between two fluidic compartments to which a differential static pressure $P$ of up to 3 bar could be applied. The streaming potential $U$ caused by the external pressure can be expressed as $\Delta U = \Delta P \varepsilon \zeta_{\text{pore}} / \eta \sigma_{\text{pore}}$. Accordingly, $\zeta_{\text{pore}}$ can be inferred from the linear relation between the applied pressure and streaming potential.

Fig. 2 shows avidin translocation experiments at varying solution pH. 40 nM avidin solution was added to a solution chamber on one side of the membrane, which was electrically grounded. The other side was biased positively or negatively ($\pm$ 150 mV) and the trans-pore ionic current was monitored over time. Spikes in the current traces are attributed to the translocation of proteins.

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to proteins passing the pore, thereby transiently blocking the current flow.

Whereas the translocations for pH 2, 6, and 8 could be explained solely by the electrophoretic motion of the protein along the electric field across the pore, an unexpected behavior was found at pH 4 and 10. At pH 10, the negatively charged protein translocates the pore towards the negatively charged electrode. This can be explained by taking the EO flow into account. Combining the equations for the EP motion of the protein and for the velocity of the EO flow yields the effective velocity of the protein inside the nanopore:

\[ v_{\text{eff}} = \varepsilon E \left( \zeta_{\text{protein}} - \zeta_{\text{pore}} \right) / \eta. \]

Due to the high nanopore zeta potential, the EO flow dominates the movement of the protein and leads to this translocation behavior. At pH 4, \( \zeta_{\text{protein}} \) and \( \zeta_{\text{pore}} \) are comparable in magnitude. Therefore, the effective force on the protein vanishes and only diffusion contributes to the translocation process. Because of that, we observe similar translocation rates for positive and negative potentials, as shown in Figure 2.

The results clearly show that EO effects have a major influence on the translocation of moderately charged nanoobjects through nanoscale orifices [3]. For small globular objects like proteins, diffusion in a concentration gradient can also be an efficient mechanism to facilitate directed translocation. As a consequence for future nanopore experiments, we conclude that the surface charge on the inner pore walls must be carefully controlled when conducting experiments with proteins and that the charging state must be accounted for in the interpretation of protein translocation events.


Supported by the TUM Institute for Advanced Studies, the International Graduate School of Science and Engineering (IGSSE), the Nanosystems Initiative Munich (NIM) and BMBF (Grant 0312031).
Data analysis of translocation events in nanopore experiments

Daniel Pedone, Matthias Firnkes, and Ulrich Rant

Nanopores have become important tools for single molecule experiments, where information about the properties of DNA/RNA or proteins can be inferred: In aqueous electrolyte solutions, an ionic current is driven through the pore by applying a voltage between electrodes placed on either side of the pore. When a molecule traverses the pore, it generates a current pulse, that is, it transiently disturbs the ionic current flowing through the pore, which can be detected with external measurement electronics (Fig. 1A). The future use of nanopores for studies of biomolecular processes like protein-protein interactions will depend on the ability to measure short translocation events. However, due to necessary electronic filters employed in the measurement technique the extraction of meaningful information from short pulses is limited. This restricts the use of nanopores for the investigation of small molecules which cross the pore rapidly. Electronic filtering affects the measurement as it distorts the signal in two ways: (i) short pulses appear dilated and (ii) the pulse height gets reduced (Fig. 1B). As the measurement techniques have reached physical limits with respect to achievable time resolution and simultaneous noise suppression, new methods are needed to recover the accurate width and height of short current pulses.

**Fig. 4:** A Representative current pulses caused by individual avidin (red) and λ-DNA (black) molecules translocating through a solid-state nanopore. The TEM image shows a nanopore in a Si3N4 membrane used for the avidin translocation. Panel B depicts the attenuation and dilation of square pulses created by a function generator when filtered with a 7-pole \( f_c = 10\text{kHz} \) Bessel filter. The nominal pulse height was 100 pA, the nominal widths were varied from 5 to 120 μs.

Here we propose two methods to analyze both the width and height of current pulses that enable us to collect meaningful data beyond the time constraints imposed by the filter electronics [1].

**Pulse width analysis.** While the individual criteria to identify the pulse-start and pulse-stop points are defined slightly differently by various authors, a common practice is to use the last data point before the current drops below the baseline as the start point and the first data point when the current returns to the baseline value as the stop point. We propose a modified way to measure the pulse width: using the same start point, the stop point is chosen to be the last (or only) local minimum of the pulse before the signal starts to return to the baseline. This criterion is based on the fact that the onset of the rising edge actually marks the end of the real pulse: When the real pulse ends, it creates an impetus for the sys-

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Pulse height analysis. Two methods are mostly used in the literature to evaluate the pulse height: some authors average over the current values within the pulse plateau to obtain a mean value, while others simply take the extremal value between the start and stop points as the pulse plateaus. The pulse height is then obtained by computing $h = I_{\text{base}} - I_{\text{plateau}}$. We devised a method to analyze the pulse height from the slope of the falling edge. We found a linear relationship between the height $h_{\text{nom}}$ and the slope $s$ of the falling edge of square pulses created by a function generator after filtering: $h_{\text{nom}} = \alpha \cdot s$. The factor $\alpha$ remains constant for $w_{\text{nom}} > 30 \mu s$ ($\alpha \approx 0.027 \mu s^{-1}$). Hence, we may infer a calibrated range, i.e., where $\alpha = \text{const.}$, for which the true pulse height can be calculated from the measured slope $s$. This means that the pulse height can be deduced from the edge slope even if the pulse is significantly shorter than $f_c^{-1}$ and the pulse has not reached its full height (developed a plateau).

![Pulse height histograms of an avidin translocation experiment. The same data set was analyzed with the conventional (blue columns) and slope method (red columns).](image)

We applied the slope method for the analysis of real protein translocation data (Fig. 1A). When analyzing the pulse height of $\sim 30,000$ translocations using the conventional method, we obtain a histogram depicted in Figure 2. The histogram peaks at a blockade current level of 50 pA which deviates significantly from the expected value of $h_{\text{ref}} = 105 \text{ pA}$, which is inferred from translocation events where avidin unspecifically adsorbed for several milliseconds to the inner pore wall (pulses reached their full height, i.e. developed a plateau). Alternatively, when we use the slope method, we obtain a distribution of current blockades which features a maximum at 91 pA, i.e., close to the expected $h_{\text{ref}}$ value. Thus, we conclude that the slope method is more suitable to recover the real height of attenuated current pulses. It enables the researcher to identify correlations and similarities among pulses in a data set which, in case of the protein measurement at hand, could have been falsely assigned to originate from distinct phenomena otherwise.

A pore-cavity-pore (PCP) device to trap and investigate single nano-scale objects in femto-liter compartments: Confined diffusion and narrow escape

Daniel Pedone, Martin Langecker, and Ulrich Rant

Because most of molecular biology operates inside cellular compartments or involves the trafficking of molecules across domain boundaries, confined diffusion has received a lot of attention. A fundamentally important problem arising in that context is the escape of nano objects from a micro-domain through a small opening (narrow escape problem).

Here we introduce an electrically addressable nano-fluidic silicon device which consists of two stacked nanopores forming the in- and outlets to a pyramidal cavity of micrometer dimensions, i.e. femto liter volume (Fig.1a).

The fabrication process involves four steps: first, the ‘SiN-pore’ is created by electron beam lithography and reactive ion etching (RIE) in the bottom SiN layer (Fig.1a&c). Second, a large square window is opened by optical lithography and RIE in the top SiN layer. Third, the pyramidal cavity above the SiN-pore is created by wet chemical etching; the resulting cavity size can be tuned by adjusting the etching time. In the final step, the Si chip is etched from the top side in a feedback controlled manner until the pyramid apex opposite from the SiN pore becomes truncated (Fig.1c). This way, the second nanopore, the ‘Si-pore’, is formed (Fig.1b). Figure 1c shows a schematic of the measurement setup. The silicon chip separates two compartments filled with aqueous electrolyte solution, each containing a Ag/AgCl electrode. The cavity interior is observed with an inverted epi-fluorescence optical microscope through the transparent SiN membrane, while an opaque layer of bulk Si (~3µm thick) blocks light from the other side of the chip. The cavity is electrically addressed by applying potentials across the PCP structure. After electrophoretically loading the cavity with a single fluorescently labelled polystyrene particle (74 nm diameter), the potential across the PCP device is set to zero. In this state, the particle’s movement is solely governed by Brownian motion. The fluorescence intensity emitted by

Fig. 1:a) 3D representation of the PCP device showing the pyramidal cavity in the silicon chip, the SiN-pore in the base center (bottom side), and the rectangular Si-pore on the top side. b) Transmission electron micrograph showing the rectangular Si-pore and the circular (smaller) SiN-pore at a time. c) Measurement setup. The chip separates two electrolyte compartments, each containing a Ag/AgCl electrode. The optical detection volume encompasses the cavity lumen, but not the upper fluidic compartment, which is blocked by the opaque silicon.
the nano-particle is observed over time to follow its trace within the cavity (Fig. 2a). To characterize the diffusive motion of a particle inside the cavity, we analyzed the mean square displacement (MSD): \[ MSD(n\Delta T) = \sum_{i=0}^{N} (r_{i+nT} - r_i)^2 / (N+1). \]
Here, \( r(t) = [x(n\Delta T), y(n\Delta T)] \) are the spatial coordinates of the particle after each data acquisition interval \( \Delta T \), and \( N \) is the total number of acquisitions (Fig. 2c).

On short time scales, the MSD increases linearly with time, as expected for free diffusion. On long time scales, however, we find that the MSD saturates at a finite value, as expected for confined diffusion. By fitting the MSD with a theoretical model we can derive the particle’s diffusion coefficient, which agrees reasonably well to the diffusion coefficient determined by dynamic light scattering. The characteristic confinement length \( \ell \) can be directly related to the pyramidal shape and size of the cavity. We found excellent agreement between these results and the MSD from Monte-Carlo simulations of a 3D random walk (Fig. 2b).

Using the PCP device we investigated the escape of single ‘Brownian’ nanoparticles (Fig. 2d) from a micro-domain (cavity) across the entropy barrier of a narrow aperture (Si-pore). The population probability is obtained by integrating the \( \tau_{\text{escape}} \) distribution from 100 escape experiments. Mean escape times \( \tilde{\tau}_{\text{escape}} \) are analyzed from single exponential fits to the population probability: \( \tilde{\tau}_{\text{escape}}^{74\text{nm}} = 8.1s \pm 0.5s; \tilde{\tau}_{\text{escape}}^{34\text{nm}} = 1.5s \pm 0.5s \) and were found to be in good agreement with analytical solutions and Monte-Carlo simulations.

Because of the addressability of the PCP device, the entropy barrier effect of the narrow apertures may be electrically amplified or attenuated. Moreover, the device dimensions are scalable in a straightforward way so that pores with diameters < 10 nm and cavity volumes in the attoliter range can be achieved easily. Thus future applications for the PCP device may include studies of confined (bio)polymers and their escape behavior or single molecule chemistry in crowded and confined environments.

Fig. 2: a) Single particle trajectory of a fluorescent 74 nm particle confined to the cavity of the PCP device recorded for 13 s with a sampling rate of 80 Hz. b) Simulated random walk trajectory within the cavity until, finally, the particle escapes. c) Mean square displacement (MSD) evaluated for particles confined in three different PCP devices with varying cavity volumes (red: \( V_1 = 18.7 \mu m^3 \); black: \( V_2 = 11.7 \mu m^3 \); blue: \( V_3 = 8.4 \mu m^3 \)). d) Population probability of 74 nm (blue) and 34 nm (black) nano-particles. Solid lines: \( P(t) \sim \exp(-t/\tau_{\text{escape}}) \)
Electrical manipulation of DNA conformations on surfaces observed with an electrochemical quartz crystal microbalance

Wolfgang Kaiser and Ulrich Rant

Active control of functional interfaces opens new opportunities for bio-sensing applications. Two factors significantly contribute to the efficiency of a surface biosensor: The accessibility of surface-immobilized probe molecules to target molecules in solution (the conformation of probe molecules on the surface), and electrostatic interactions between the supporting surface and charged target molecules. By controlling one or both properties binding efficiencies may be enhanced or suppressed.

We focus here on the conformation- and elasticity-changes of low-density DNA layers caused by externally applied potentials. Furthermore, we monitor the self assembly of a DNA layer, the surface passivation by a mercaptohexanol (MCH) self-assembled monolayer (SAM), and the electrically induced partial desorption of DNA in order to produce a low-density DNA layer. All measurements were done in real-time by an electrochemical quartz crystal microbalance with simultaneous dissipation monitoring (EC-QCM-D) and compared to fluorescence measurements (data not shown).

**DNA-SAM formation.** The gold surface of a QCM was exposed to a solution of thiolated double stranded DNA (dsDNA) with 72 base-pairs. Figure 1 (left) shows the self-assembly of a dsDNA-layer in real-time by monitoring the shift in resonance frequency $\Delta f$ as well as the dissipation $\Delta D$ of the QCM. The DNA layer formation is completed after ~300 sec, as indicated by the saturation of $\Delta f$ as well as $\Delta D$. Rinsing the sensor with buffer solution to remove abundant DNA afterwards does neither affect the frequency nor the dissipation. In a second step MCH was co-adsorbed to passivate the gold electrode and to remove un-specifically bound DNA (mid part of Figure 1). Changes in $\Delta f$ and $\Delta D$ confirm the partial replacement of DNA with MCH. Finally, a part of the DNA is electrically desorbed from the gold surface in MCH solution (Figure 1 right) by applying successive negative potentials to the gold electrode versus a platinum counter electrode.

The change of frequency and dissipation after the individual potential steps clearly indicates that the mass bound to the gold electrode (DNA) decreases (smaller $\Delta f$) and that the layer becomes more rigid (smaller $\Delta D$). The increased rigidity probably arises from a DNA layer orientation closer to the surface, since for individual strands without interactions with their nearest neighbors a configuration close to the surface is entropically favorable. The electrical desorption steps are important in order to obtain a low-density DNA layer, where

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single DNA molecules behave as individual entities and can align in the electric field without sterical interactions with neighboring strands.

**EC-QCM-D monitoring of DNA alignment in electric fields.** When applying electric potentials to the gold surface, the intrinsically negatively charged DNA molecules align in the electric field, as has been shown by scanning probe microscopy and fluorescence energy transfer techniques [1]. The current results confirm that the electric field effect is also apparent in the QCM-D signals: Figure 2 shows how the resonance frequency increases when sweeping the potential positively, while the dissipation decreases concurrently.

This can be rationalized by the following interpretation: at negative potentials the DNA strands extend away from the surface and a large amount of trapped water (mass) moves together with the DNA layer, which is comparably flexible in this state. As a result, the quartz oscillation is strongly damped which results in highly negative frequency shifts and pronounced dissipation (which is a measure for the layer rigidity). At positive potentials the DNA is electrostatically pulled towards the surface; less water is forced to move with the DNA layer, which is also quite rigid. Thus, the oscillation frequency is higher and the dissipation is lower. A schematic drawing for attractive and repulsive potentials is presented in Figure 3 A and B, respectively.

In conclusion, the electrically induced conformation switching of DNA molecules tethered to a QCM surface directly results in a significant oscillation frequency change of ~ 4 Hz. In addition, the altered energy dissipation of the system indicates that a DNA layer compressed onto a positively charged surface is considerably more rigid than a ‘brush-like’ DNA layer extending from a negatively charged surface.


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Supported by the TUM International Graduate School of Science and Engineering, the TUM Institute for Advanced Studies, and LOT Oriel / Q-Sense
Kinetics of target-binding to surface-tethered probes studied by the \textit{switch}DNA technique

Makiko Maruyama, Gerhard Abstreiter, and Ulrich Rant

The overall detection performance of a surface biosensor depends on the transport of target molecules from solution to the sensor and the reaction rate of the targets with the surface-immobilized probes. A limitation due to mass transport occurs when the target-probe reaction is faster than the target molecule diffusion. A simple and effective approach to overcome this limitation is convective analyte flow across the sensor surface. In this work, we report binding kinetics studied with the \textit{switch}DNA technique for different target molecules: DNA oligonucleotides, IgG antibodies, and streptavidin, focusing on the influence of analyte flow. The orientation of DNA molecules tethered to a gold surface can be manipulated effectively by electric fields and the fluorescence intensity of a dye at the distal end of the DNA strands is used to monitor the conformation change. The \textit{switch}DNA technique allows for the detection of DNA \cite{1} as well as protein targets \cite{2} in real-time. Target-probe binding is monitored by changes in the switching behavior of the probe DNA layer, i.e., by a change of the fluorescence modulation, $\Delta F$.

The rate at which probe molecules capture targets is either limited by mass transport of targets to the sensor area or by the chemical reaction between probes and targets. When no analyte flow is applied and the chemical reaction is fast, kinetics are dominated by the diffusion of targets to the sensor surface (diffusion limited kinetics). Under continuous analyte flow, the dominant contribution to the sensor kinetics can be assessed by the Damköhler number, which denotes the ratio of reaction to diffusion rate \cite{3}. If the probe-target reaction is faster than the target’s diffusion ($Da > 1$), the target depletion zone above the sensor surface remains constant in size. In this case, the target surface coverage increases linearly with time due to a constant diffusive flux across the depletion zone (convection-diffusion limited kinetics). In contrast, a system with $Da < 1$ is reaction limited. In this case, targets molecules are transported efficiently to the sensor surface and are always present in abundance (reaction limited kinetics).

Fig. 1 compares real-time binding measurements with different 24mer DNA targets and IgG antibody as well as streptavidin protein targets under conditions of stopped flow and applied flow, respectively. The applied analyte flow does not affect the observed DNA binding kinetics, which shows that the system operates in the reaction limited regime. This is also reflected by a low $Da$ number $\sim 0.03 << 1$, which results from the moderate reaction rate of DNA-DNA interactions. The fit of the reaction kinetics model to the data gives an excellent match with a residual square value $> 0.97$. In contrast to literature reports for other surface biosensors, the forward rate constants obtained on the surface ($k_{on} > 10^5$ M$^{-1}$s$^{-1}$) are remarkably high and correspond to values measured in solution.

The protein binding experiments, on the contrary, exhibit significantly accelerated kinetics when applying flow. Closer inspection reveals that the kinetics under continuous flow do neither agree with diffusion limited nor with reaction limited models, but may be described with a convection-diffusion kinetics model. The results are consistent with expectations from the calculated $Da$ number $\sim 5 > 1$, which indicates that the system operates in the convection-diffusion regime.

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Supported by the International Graduate School Materials Science of Complex Interfaces (CompInt), the TUM Institute for Advanced Studies, and Fujitsu Laboratories Ltd.

In conclusion, we have investigated the kinetics of target binding to surface-tethered probes by the switchDNA technique. The influences of analyte flow are studied for DNA (with moderate $k_{on}$) and protein (with high $k_{on}$) sensing. The hybridization of 24bp DNA on the switchDNA sensor proceeds with remarkable efficiency, which matches hybridization rates found in solution. The results indicate that the sensor predominantly operates in the reaction-limited regime for the DNA binding experiments. On the other hand, mass transport limited kinetics are found to govern the binding of IgG and streptavidin to surface-tethered biotinylated probes due to their fast probe-target chemical reaction. In this case, the binding kinetics are strongly enhanced by applying an analyte flow.


Platform for in situ real-time measurement of protein-induced conformational changes of DNA

Philipp S. Spuhler¹, Jelena Knežević², Ayça Yalçın¹, Qiuye Bao³, Erika Pringsheim², Peter Dröge³, Ulrich Rant², and M. Selim Ünlü¹

Understanding conformational dynamics and orientation of biomolecules is critical in determining their function. Regulation of gene expression, for example, in both prokaryotes and eukaryotes involves formation of specialized nucleoprotein structures, or snups. Within these complexes, distant segments of DNA are brought into close proximity to each other and, frequently, protein-induced DNA bends or kinks are formed. Understanding the DNA conformational changes and forces responsible for bending DNA to form specific complexes are thus of considerable biological significance.

Here, we introduce a platform for in situ real-time measurement of protein-induced conformational changes on immobilized DNA [1]. We combine electrically induced orientation of surface-bound dsDNA probes with an optical technique to measure the kinetics of DNA conformational changes [2]. The principle for measurement of orientation and conformation of surface-bound dsDNA is depicted in Fig. 1.

The efficiency of nonradiative energy transfer (ET) between the fluorophores and surface plasmons in gold follows a distance dependency $ET \propto d^{-3}$ (d is the distance between the fluorophore and the gold surface) and allows real-time observation of the emitter-gold distance through the fluorescence intensity. The sequence-specific Escherichia coli integration host factor is utilized to demonstrate protein-induced bending upon binding of integration host factor to dsDNA probes [3].

We show that the electrical control of probe orientation is a necessary feature to discern conformational changes of surface bound dsDNA probes in this type of measurement (Fig. 2).

![Fig. 1: Application of a negative potential repels the DNA probe from the surface and results in high fluorescence intensity because of inefficient ET between Cy3 and the gold electrode, whereas positive potentials pull the DNA probes to the surface, resulting in reduced fluorescent intensity.](image_url)

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Furthermore, we present the influence of IHF binding sequence position in the DNA probes on the sensor performance. The effects of probe surface density on binding/bending kinetics are investigated (data not shown).

![Graph showing normalized fluorescence of Cy3-labeled oligonucleotide probes measured prior to and after IHF binding while the applied voltage potential is swept from negative (oligonucleotides oriented in the standing position) to positive (oligonucleotides oriented in the lying position). Probes (H'-High, H'-Mid, H'-Low, and Control) were immobilized on separate gold electrodes. IHF-induced conformation change is readily identified when oligonucleotides are oriented in the standing position with a negative applied potential. The IHF concentration is 50 nM.](image)

**Fig. 2:** Normalized fluorescence of Cy3-labeled oligonucleotide probes measured prior to and after IHF binding while the applied voltage potential is swept from negative (oligonucleotides oriented in the standing position) to positive (oligonucleotides oriented in the lying position). Probes (H'-High, H'-Mid, H'-Low, and Control) were immobilized on separate gold electrodes. IHF-induced conformation change is readily identified when oligonucleotides are oriented in the standing position with a negative applied potential. The IHF concentration is 50 nM.

We demonstrated the ability to discern differences in DNA conformational changes resulting from sequence-specific protein binding. The measured concentration dependence of protein binding agrees with previously published values (data not shown). To address a high dependence of the binding kinetics on the probe surface density, we describe a method to conduct multiple binding experiments to a surface with high repeatability in the binding kinetics. The platform therefore offers a comprehensive approach to investigate protein-induced conformational changes in immobilized dsDNA. Additionally, the platform can accommodate individual spots of microarrayed dsDNA on individually controlled, lithographically designed electrodes, making it amenable for use as a high throughput assay.

Conformation of end-tethered DNA molecules on gold surfaces: Influences of applied electric potential, electrolyte screening, and temperature

Wolfgang Kaiser and Ulrich Rant

DNA monolayers anchored on solid supports are a paradigm of functional molecular interfaces with multifaceted physical and chemical properties. Although they have already found their way into important applications like DNA microarray technology, their full potential has not been tapped yet. Owing to their charged nature, surface-tethered DNA layers can be manipulated efficiently through the application of electric fields and by this means can be turned from passive layers into actively controlled nano-bio interfaces with improved functionality.

Here we systematically study the influence of externally applied surface potential and the consequences of varying solution salinity as well as temperature on the conformation of DNA layers comprising 72mer nucleotides [1]. Special care has been taken to prepare ultra-low-density DNA layers so that interactions between DNA molecules on the surface (crowding) can be excluded. The extension of the upper DNA from the surface is inferred from quantitative fluorescence energy transfer measurements.

In solutions of high ionic strength (cf. Figure 1A, 1B and 2A), where electric interactions are almost completely suppressed due to the presence of high concentrations of ions in solution, double-stranded (ds) DNA and single-stranded (ss) DNA adopt very different conformations on the surface.

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While single strands take on a compact state, double strands stand on the surface, most likely due to steric interactions of the surface-near part of the rigid helix with the underlying alkanethiol layer. The molecular conformations are extremely “robust” in that neither variations in the surface potential nor temperature changes substantially affect the molecules. In an intermediate salinity range (cf. Figure 1C-F, 2B), electric fields emanating from the charged surface interact with the surface-proximal parts of the DNA oligonucleotides. It is possible to efficiently manipulate the DNA conformation by the application of bias potentials to the supporting metal electrodes. The effectiveness of the induced conformation change is higher for double- than for single-stranded oligonucleotides because the former behave like rigid rods. At low ionic strength (cf. Figure 1G, 1H and 2C), the structural properties of ss-DNA change dramatically. Due to electrostatic repulsions between charged sites along the DNA backbone, the electrostatic contribution to the persistence length strongly increases; consequently, the single strands stiffen and become rod-like. In this state the extension of ss-DNA from the surface may be manipulated very efficiently by electric fields.

The conformation temperature dependence of single- and double-stranded DNA molecules is very different, except at high salt concentrations, for which both ss-DNA and ds-DNA are fairly insensitive to temperature changes. While ds-DNA layers generally retain this robustness against temperature changes up to the duplex melting temperature, ss-DNA layers exhibit profound conformation changes upon temperature variations. In an intermediate salinity range, the conformation changes can be attributed mainly to the melting of intra-strand base pairs, i.e., the melting of short duplexes that keep the otherwise single-stranded molecules partly folded. A very intriguing behavior was revealed in low-salinity solutions: ss-DNA stretches out substantially with increasing temperatures. This phenomenon is probably caused by the entropy-driven release of counterions from ss-DNA, the disruption of base stacking at elevated temperature, or a combination of both. As a consequence, the effective charge density along the DNA polymer increases and leads to an enhanced electrostatic stiffening/stretching of the otherwise flexible single strands.

Fig. 2: Schematic representation of the DNA conformation on negatively and positively charged surfaces in electrolyte of varying ionic strength.


Supported by the TUM International Graduate School for Science and Engineering (IGSSE), the TUM Institute for Advanced Studies, and Fujitsu Laboratories Ltd.
Formation and separation of gold nanoparticle dimers with a protein linker

Vera B. Zon, Matthias Sachsenhauser, and Ulrich Rant

The colloids of noble metals attract substantial interest in recent years because of their exceptional optical properties with applications in optoelectronics, surface-enhanced Raman scattering, fluorescence or IR absorbance spectroscopy. The optical properties of the metal nanoparticles (NPs) are governed by localized surface plasmon resonances (LSPRs), which are collective oscillations of the particles’ conduction electrons and the incident light. The coupling of LSPRs between closely located colloidal metal particles is particularly interesting, because it generates massive enhancements of the electric field strength, so called ‘hot spots’, which may be used for surface-enhanced spectroscopy or engineered for energy-transfer constructs.

The formation of nanoparticle constructs can be achieved via different routes, for instance using ‘conventional’ chemical conjugation chemistry or DNA mediated assembly. We study the protein-induced aggregation of biotin-modified gold nanoparticles, which occurs due to specific interaction between streptavidin and biotin [1]. This process can be adjusted to form either larger or smaller structures by tuning, e.g., pH value of the solution or by controlling the NP:streptavidin stoichiometry. In the present study we focus on the preparation of NP dimers; the principle is illustrated in Fig.1.

Here we report the controlled formation and separation of gold NP dimers. The formation process has been studied with UV/VIS extinction spectroscopy, dynamic light scattering (DLS) and transmission electron microscopy (TEM). The plasmon resonance of metal NPs is very sensitive both to size and local environment of the particles. Therefore, formation of NP aggregates usually leads to a shift or broadening of the LSPR peak. Moreover, the dimerization (or multi-merization) of particles can also be followed by DLS where a shift in the NP size distribution occurs upon cross-linking with streptavidin.

The synthesis procedure was carried out at near neutral pH values and involved three steps: (i) functionalization of gold NPs with biotin, (ii) NP passivation with thiolated PEG, and (iii) introduction of streptavidin to link the biotin-modified NPs. Commercially available citrate-coated 20 nm gold NPs were functionalized with biotin by means of ligand exchange with a biotin-modified PEG disulphide: NPs and biotin-PEG were mixed and allowed to react for 12 hours. The second step in the synthesis was required to decrease the absolute value of the ζ-potential of the NPs (not shown here), since both the NPs and streptavidin are negatively charged in this pH range, which affects their reactivity. The remaining citrate was replaced with thiolated PEG (HS-EG₆-CH₃) in

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another ligand exchange reaction. The reaction time of 2 hours was critical in this case; if it was too short, the NPs would still be highly negatively charged, if it was too long, it lead to the replacement of biotin-modified PEG on the particle surface by the more reactive thiolated PEG molecules. In the last step controlled-aggregation of the NPs could be performed by adding the cross-linking protein streptavidin.

The size of the agglomerates was measured by DLS. As soon as binding between biotin and streptavidin occurred, we observed a shift to greater diameters in the size distribution, as well as a wavelength shift of the plasmon peak maximum in the extinction spectra. These parameters were monitored to optimize the conditions to achieve a well-defined NP linkage; in particular, we examined the influence of streptavidin and NP concentrations, the density of the biotin layer on the particle surface, and buffer and pH effects. The protein linked samples exhibited a peak at 40 nm in the DLS size distribution, which roughly corresponds to the expectation of two linked 20 nm particles.

Purification was necessary to enrich the fraction of protein linked gold NP dimers. Size exclusion chromatography (SEC) was used to purify the linked dimers from reaction precursors and particle monomers [2]. A typical chromatogram is presented in Fig. 2. In the protein linked sample (red curve) a shoulder appears at the otherwise sharp 20 nm gold NP monomer peak (black). Since shorter retention times correspond to greater sizes of the eluted particles, this shoulder is attributed to NP-dimers. Moreover, extinction spectra of the ‘dimer’ shoulder are red-shifted (see Fig 2b), as expected for coupled NPs. The ‘dimer’ fraction was collected and examined by TEM; a typical image is presented in Fig. 2 (a) confirming the presence of NP dimers.

In conclusion, we have demonstrated a facile route to form gold nanoparticle dimers, and their purification by size exclusion chromatography.


Fig. 2: Size exclusion chromatogram for monomeric 20 nm gold NPs (black curve) and after addition of the streptavidin linker (red curve). Inset (a) TEM image depicting dimers from the collected fraction; inset (b) extinction spectra showing the red-shifted LSPR of the dimer fraction for t<13 min.
Photoinduced growth of silver nanoparticles on Cy5-modified DNA

Vera B. Zon and Ulrich Rant

The colloids of noble metals have proved exceptional properties in such areas as catalysis, optoelectronics, surface-enhanced Raman scattering, fluorescence and IR absorbance spectroscopy. The optical properties of metal nanoparticles (NPs) are governed by localized surface plasmon resonances (LSPRs), which result from the interaction of the conduction electrons of the particles with the incident light. Furthermore, the DNA-programmed assembly of NPs is especially intriguing, because particles can be deterministically arranged on the nano-scale. Two routes can be considered to prepare NP-DNA conjugates as building blocks for DNA-templated NP structures: (i) conjugation of pre-formed NPs with DNA through suitable coupling chemistry or (ii) direct growth of metal particles on predesigned DNA molecules. An advantage of the latter approach is that the DNA : NP stoichiometry can be well controlled which avoids the necessity for cumbersome purification procedures. Methods to nucleate and grow silver particles directly on a DNA molecule involve either modification of DNA with aldehyde moieties (Tollens reaction), introduction of reducing agents, or illumination with UV light. A particularly interesting method has been put forward by Hammond et al. [1], which is based on the principle of spectrally sensitized photography. The process involves the photoinduced reduction of Ag⁺ ions by optically excited dye molecules [2] and has been used for the sensitive detection of dye-labeled DNA.

Based on these reports, we investigated the growth of silver NPs on 48 base-pair double-stranded Cy5-labeled DNA by photo-induced reaction. DNA solutions were irradiated with light from a Kr ion laser (647 nm, ca. 70 mW) at the Cy5 absorption maximum until >90% of the dye molecules were photobleached. As a silver source and reducing/capping agent, the solutions contained AgNO₃ and sodium citrate in 1 mM concentrations, respectively. The nanoparticle growth process was studied by UV/VIS extinction spectroscopy, dynamic light scattering (DLS), transmission electron microscopy (TEM), and dark-field optical microscopy.

Figure 1 shows the photoinduced formation of Ag-NPs as evidenced by UV/VIS extinction spectroscopy. As a result of the intense laser irradiation and concomitant photo-bleaching, the Cy5 peak decreases to <10% of its initial value. At the same time, we observe the occurrence of a peak at ~420 nm which is associated with LSPR of silver nanoparticles. Although this peak appears concurrently with the Cy5 photo-bleaching, it is of minor magnitude initially and keeps on growing over a time period of several days. Moreover, the growth rate was found to strongly depend on the concentration of silver ions in solution.

![Extinction spectra showing photobleaching of Cy5 under laser irradiation and the LSPR development over 4 days.](image)

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DLS and TEM analysis (not shown here) confirm the presence of metallic nanoparticles. However, while DLS results reveal a size distribution which peaks at ca. 35 nm, roughly spherical particles with a most frequent diameter of ca. 15 nm are found in the TEM images. A possible reason for this discrepancy could be that TEM visualizes the metallic core of the particles, whereas DLS measures their hydrodynamic diameter, which can be enhanced due to the presence of organic adsorbates on the particle surface.

In order to test whether the formed NPs are associated with DNA molecules, centrifugation experiments were performed. Figure 2 shows the extinction spectra of a NP sample grown by photoinduced DNA metallization (Fig. 2 a) and a reference sample where DNA was mixed with commercially obtained 20 nm silver NPs (Fig. 2 b). Both, DNA (260 nm) and NP (~400 nm) extinction peaks are detected before centrifugation (red curves). After centrifugation the supernatant was removed and the extinction of the re-dissolved nanoparticle precipitate measured again (blue curves). Now, the reference sample is void of DNA and only the NP LSPR peak is visible in Fig. 2b. In contrast, the sample with the metallized DNA (Fig. 2a) exhibits a distinct peak at 260 nm confirming the presence of DNA. A possible explanation for this finding is that DNA sticks out of metal nanoparticles which nucleated at the Cy5 labeled end of the oligonucleotides. However, the alternate possibility of DNA wrapping around the Ag NPs and capping them during the growth process cannot be excluded at this point. Further experiments attempting the DNA mediated assembly of NPs are required to resolve this issue.

In conclusion, we have examined a photo-induced synthesis procedure to grow silver NPs on dye-labeled DNA oligonucleotides. The obtained colloids are mostly spherical in shape (ca. 15 nm in diameter), exhibit a typical plasmon resonance in their extinction spectra, and are associated with DNA.


Fig. 2: Extinction spectra normalized to the LSPR maxima of the solutions before centrifugation and re-dispersed precipitates after centrifugation: (a) a sample grown by photo-induced metallization of Cy5-labeled DNA and (b) a reference sample, which is mixture of silver colloids and DNA.
On-chip functionalization of carbon nanotubes by the photosystem I

S. M. Kaniber, M. Brandstetter, F. C. Simmel, I. Carmeli, and A.W. Holleitner

Photosynthesis in plants and bacteria is driven by molecular electronic complexes, such as the photosystem I (PSI). Photoexcitation of PSI causes an electron transfer through a series of redox reactions from the chlorophyll special pair P700, as a primary electron donor, to a final electron acceptor, ∼6 nm away from the oxidant P700⁺. Here, we used carbon nanotubes (CNTs) as nanoscale electrical wires to electrically contact the PSI [1,2]. We compared the optoelectronic properties of the CNT-PSI hybrids for three different on-chip functionalization strategies [3]. The PSI was bound to the CNTs via covalent, hydrogen, or electrostatic bonds.

For device fabrication, individual carboxylated CNTs are first deposited on an insulating SiO₂ substrate and then contacted with Pd source-drain electrodes by e-beam lithography. In the first chemical route, selectively generated cysteins on the lumenal side of the PSI reacted with maleimide functionalized CNTs. AFM images before and after functionalization of the CNTs with PSI provide a first indication that PSI proteins are attached to the sidewalls of the CNTs due to covalent bonding between the CNT and the PSI. Hence, the electron transfer path in PSI is expected to be perpendicular to the CNT (sketch in Figure 2a).

Filled squares in Figure 2a depict the wavelength dependence of the photoconductance of sample A, which was fabricated by the first chemical route. The photoconductance showed an enhanced signal at 680 and 360 nm. The findings are consistent with the absorbance data of the PSI (solid line in Figure 2a). Importantly, the enhanced Gₚₛₛₚ below 700 nm can be attributed to an effect of the PSI. This was proven by measurements on sample B (Figure 2b). Here, the photoconductance of a CNT bundle before and after on-chip functionalization.

Fig. 1: (a-d) Wavelength dependence of the photoconductance of sample A, B, C and D which are fabricated according to different on-chip functionalization strategies. The electron transfer path has different orientations with respect to the CNTs’ axis [3].

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tion with PSI by covalent bonding was compared. The maxima at 680 nm and at ~360 nm only appear when the PSI is attached to the CNTs. Above 700 nm, the PSI does not absorb light, and in turn, the magnitude of the photoconductance before and after functionalization with PSI is the same within the experimental error. Consequently, the two photoconductance maxima around 760 and 845 nm and also the maximum at 825 nm in Figure 2a are interpreted to result from electron-hole dynamics only within the contacted CNTs [2,3]. These findings are consistent with resonances assigned to the E_{22} transition in semiconducting CNTs with a diameter in the range 1-2 nm [4]. The origin of the enhanced photoconductance at the resonances of the PSI can be attributed to an electron or energy transfer between the excited PSI and the CNTs.

In the second chemical route, the electrically contacted CNT bundle of sample C was functionalized with ethylenediamine before the reaction with the PSI. AFM measurements proved that the PSI is adsorbed to the CNTs. This can be explained by hydrogen bond formation or electrostatic interactions between the PSI and the amino groups on the CNTs. In both cases, the PSI was preferentially orientated with its electron transfer path perpendicular to the amine modified CNTs (sketch in Figure 2c). The photoconductance of sample C showed a peak at 680 nm (Figure 2c) which again can be assigned to a charge or energy transfer between the PSI and the CNT.

In the third chemical route (sample D), the functionalization of the CNTs with PSI was enabled by electrostatic forces between negatively charged terminal groups on the CNTs and positively charged regions on the luminal and stromal side of the PSI. AFM measurements demonstrated a partial adsorption of PSI to the CNTs. The wavelength dependent photoconductance of sample D (squares in Figure 2d), however, showed only resonances above 700 nm which are caused by electron-hole dynamics in semiconducting CNTs. Most importantly, there was no photoconductance resonance at 680 nm (triangle in Figure 2d), a fact which can be explained by a less efficient adsorption of the PSI to the CNTs and by considering the mechanism of adsorption of the PSI on the CNTs. Electrostatic interactions between negatively charged COO- groups and the PSI favor a parallel orientation of the PSI’s electron transfer path with respect to the CNT. For this alignment the photogenerated dipole is parallel to the CNTs’ axis and the proposed mechanism of energy transfer is less efficient. In addition, a possible charge transfer between the PSI and the CNT is suppressed in this configuration [2,3].

Supported by the DFG via Grant HO3324 and the Nanosystems Initiative Munich (NIM).
3. Research Funding and Collaborations

Many of our research projects have benefited from very fruitful collaborations with external groups either via joint projects, individual collaborations, exchange programs, or through direct interaction with visitors. The major external collaborations are based on joint projects which are financially supported by different organizations. The total amount of external funding exceeded 3 Mio € in 2009.

External Funding

1. Bundesministerium für Bildung und Forschung (BMBF)
   - Kompetenzzentrum Nanop: „Anwendung von Nanostrukturen in der Optoelektronik“
   - NOSE (Nanotechnologie für die optische Spektroskopie)
   - Verbundvorhaben EPR-Solar – „Ortsaufgelöste elektrisch detektierte magnetische Resonanz“
   - Nanobiotechnologie-Verbundprojekt: MultiplexLAB on NanoChip; TP 2

2. Bundesministerium für Wirtschaft und Technologie (BMWi)
   - Exist-Forschungstransfer: switchSENSE – Bio – Chip Technologie zur parallelen Detektion und Analyse von Proteinen

3. Deutsche Forschungsgemeinschaft (DFG)
   - Sonderforschungsbereich “Festkörperbasierte Quanteninformationsverarbeitung” (SFB 631)
   - Schwerpunktprogramm “Halbleiter-Spintronik“
   - Forschungskooperation “FORMEDIAN“
   - AB 35/6-1: “Silicon and germanium based magnetic semiconductors”
   - AB 35/6-2: “Spin injection and detection in silicon based heterostructures”
   - AB 35/8-1: “International collaboration in chemistry: A new molecular architecture for organic electronics”
   - BR 1585/5-1: “Quantum properties of dopants for silicon nanospintronics“
   - BR 1585/6-1: “Nanostrukturierte Thermoelektrika: Modellsysteme und kontrollierte Synthese"
   - EI 518/2-1: ERA NANOSCI: “Nanowire arrays for multifunctional chemical sensors”
   - HO 3324//2-1: “Quantum transport at the molecular scale”
   - HO 3324/2-2: “Molecular circuits for optoelectronics utilizing carbon nanotubes”
   - HO 3324/4-1: “Optoelectronic detection of non-equilibrium spin transport in the mesoscopic regime in semiconductor nanostructures”
   - HO 3324/4-2: “Optically induced non-equilibrium spin transport in mesoscopic semiconductor circuits”
   - VO 483/5-2: “Microscopic theory of spin-splittings and ballistic spin currents in semiconductor“
4. **Excellence Initiative of Federal and State Governments**
   - Cluster of Excellence „Nanosystems Initiative Munich“ (NIM, 7 projects)
   - TUM International Graduate School of Science and Engineering (IGSSE)
   - TUM Institute for Advanced Study (IAS)
   - TUM “Towards bionic solar cells”

5. **Bayerisch-Kalifornisches Hochschulzentren – BaCaTeC**
   “Scanning force microscopy on hydrogen terminated diamond surfaces”

6. **DAAD**
   - Partnerschaften mit Japan und Korea

7. **Alexander von Humboldt-Stiftung**

8. **Graduiertenförderung:**
   Elitenetzwerk Bayern
   Int. Graduate School “Materials Science of Complex Interfaces (CompInt)”
   Karl-Max von Bauernfeind-Verein, TUM
   Universität Bayern eV

9. **European Union**
   - DOTSENSE (Group III-nitride quantum dots as optical transducers for chemical sensor)
   - DREAMS (Diamond to Retina Artificial Micro-interface Structures)
   - Marie Curie Excellence Grant: SENFED (Semiconductor Nanowires and their Field Effect Devices) EXT-CT-2006-042721
   - Marie Curie FP7 Peoples – International Reintegration Grant (IRG): “Self-assembled growth of III–V Semiconductor Nanowires on Si for Future Photonic and High Electron Mobility Applications”
   - Marie Curie “III-V NWs on Si”
   - MATCOM (Materials and Interfaces for Energy Conversion and Storage)
   - NEMIS (New Mid-Infrared Sensors)
   - SUBTUNE (Widely Tunable VCSEL using sub wavelength gratings)

10. **FWF - Fonds zur Förderung der wissenschaftlichen Forschung, Wien, Austria**
    SFB F025 “InfraRed optical nanostructures”
11. Industry

– Fujitsu Laboratories Ltd.: “SwitchDNA Biosensor”
– “IPB: “Polykristalline Dünnschicht”
– Rhode & Schwarz, München, Germany: “Zero-Bias Schottky-Diodes“
– Siemens AG, München: “Defekte in CdTe“
– Siemens AG, Erlangen: “Electrodes for thermionic converter“
– Vertilas, Garching, Germany: “Langwellige VCSEL“
## International collaborations

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<td>Jocelyn Achard</td>
<td>Centre National de la Recherche Scientifique, Paris, France</td>
<td>High purity single crystalline diamond substrates</td>
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<tr>
<td>Yasuhiko Arakawa</td>
<td>University of Tokyo, Japan</td>
<td>Optical spectroscopy of quantum dots</td>
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<tr>
<td>Günther Bauer</td>
<td>Universität Linz, Austria</td>
<td>Optical properties of quantum dot lattices X-ray analysis of quantum dots</td>
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<tr>
<td>Philippe Bergonzo, Milos Nesladek</td>
<td>Commissariat à l’Energie Atomique, Saclay, France</td>
<td>Neurons on diamond surfaces</td>
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<tr>
<td>Tim Boykin</td>
<td>University of Utah, USA</td>
<td>Tight binding methods</td>
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<td>Mark Brongersma</td>
<td>Stanford University, USA</td>
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<td>Glenn Burley</td>
<td>University of Leicester, UK</td>
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<td>Etienne Bustarret</td>
<td>Centre National de la Recherche Scientifique, Grenoble, France</td>
<td>Low temperature electronic transport in diamond, III-nitrides on diamond</td>
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<td>Jens Buus</td>
<td>Gayton Photonics, Gayton, UK</td>
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<td>Michel Calame</td>
<td>Universität Basel, Switzerland</td>
<td>Optoelectronics of nanoparticle arrays</td>
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<tr>
<td>Fernando Calle</td>
<td>Universidad Politècnica de Madrid, Madrid, Spain</td>
<td>AlGaN/GaN electrolyte gate FETs, GaN MEMS and NEMS</td>
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<tr>
<td>Itai Carmeli</td>
<td>Dep. of Physical Chemistry and Center for Nanoscience Tel-Aviv University, Israel</td>
<td>Optoelectronics with photosynthetic reaction center (PSI)</td>
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<td>Yves Chabal</td>
<td>Rutgers University, Piscataway, USA</td>
<td>FTIR studies of hydrogen-induced exfoliation of Ge</td>
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<td>Connie Chang-Hasnain</td>
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<td>Subwavelength gratings, long-wavelength VCSELs</td>
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<td>Lukas Chrostowski</td>
<td>University of British Columbia, Vancouver, Canada</td>
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<td>Klaus Ensslin</td>
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<td>Lars Frandsen / Jørn Hvam</td>
<td>DTU Copenhagen, Denmark</td>
<td>Fabrication of SOI photonic crystal nanostructures</td>
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<td>Dieter M. Gruen</td>
<td>Argonne National Laboratory, USA</td>
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<tr>
<td>Ken Haenen</td>
<td>Interuniversitair Micro-Electronica Centrum, Hasselt, Belgium</td>
<td>Nanocrystalline diamond growth</td>
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<tr>
<td>Eugene E. Haller</td>
<td>University of California, Berkeley, USA</td>
<td>Isotopical engineering of Si and Ge</td>
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<td>Robert Hamers</td>
<td>University of Wisconsin, Wisconsin, USA</td>
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<td>Ulrich Höhenester</td>
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<td>Eduard Hulicius</td>
<td>Institute of Physics, Czech Academy of Sciences, Prague, Czech Republic</td>
<td>Antimonide VCSELs</td>
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<td>Kohei M. Itoh</td>
<td>Keio University, Yokohama, Japan</td>
<td>Isotopical engineering of Si and Ge</td>
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<td>Richard Jackman</td>
<td>University of College London, London, UK</td>
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<tr>
<td>Paul Koenraad</td>
<td>Technical University of Eindhoven, The Netherlands</td>
<td>Scanning tunnelling microscopy of coupled quantum dot nanostructures</td>
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<td>Sigmund Kohler</td>
<td>Instituto de Ciencia de Materiales de Madrid, CSIC</td>
<td>Ballistic photocurrents in mesoscopic structures</td>
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<td>Philomela Komninou</td>
<td>Aristotle University of Thessaloniki, Greece</td>
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<td>Wolfgang Langbein</td>
<td>University of Cardiff, UK</td>
<td>Ultrafast FWM spectroscopy of nanostructures</td>
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<td>Anders Larsson</td>
<td>Chalmers University, Gothenburg, Sweden</td>
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<td>Karin Larsson</td>
<td>University of Uppsala, Uppsala, Sweden</td>
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<td>Steve Laux</td>
<td>IBM, Yorktown Heights, USA</td>
<td>Quantum transport</td>
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<td>Kwan H. Lee</td>
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<td>Simulation of nitride based nanorods and quantum dots</td>
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<td>Peter Lodahl / Søren Stobbe</td>
<td>DTU Copenhagen, Denmark</td>
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<td>Enrique López Cabarcos</td>
<td>Universidad Complutense de Madrid, Spain</td>
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<td>Daniel Loss</td>
<td>University of Basel, Switzerland</td>
<td>Theory of carrier spin relaxation in QDs</td>
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<td>Max Migliorato</td>
<td>University of Manchester, UK</td>
<td>Electronic structure calculations in strained nanostructures</td>
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<td>Eva Monroy</td>
<td>Commissariat à l’Energie Atomique, Grenoble, France</td>
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<tr>
<td>Elias Munoz</td>
<td>Universidad Politécnica de Madrid, Madrid, Spain</td>
<td>Noise characterization of diamond transistors</td>
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<td>Mark Newton</td>
<td>University of Warwick, Warwick, UK</td>
<td>Characterization of diamond films</td>
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<td>David Norris</td>
<td>University of Minnesota, USA</td>
<td>Optical properties of colloidal QDs</td>
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<td>Vittorio Pellegrini</td>
<td>University Pisa, Italy</td>
<td>Low energy excitations in quantum wires</td>
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<td>Pierre Petroff</td>
<td>University of California, Santa Barbara, USA</td>
<td>Photonic crystal nanocavities</td>
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<td>Serge Picaud</td>
<td>Institut National de la Santé et de la Recherche Médicale, Paris, France</td>
<td>Retina cells on diamond surfaces</td>
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<td>Michael Povolotskyi, Aldo di Carlo</td>
<td>University of Rome „Tor Vergata“, Italy</td>
<td>Quantum transport, tight binding methods and nextnano³ simulations</td>
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<td>Joan Ramon Morante</td>
<td>Universitat de Barcelona, Spain</td>
<td>III-nitride and metal-oxide nanostructures</td>
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<td>Jordi Arbiol</td>
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<td>High resolution transmission electron microscopy of nanowires</td>
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<td>Andrew Ramsay</td>
<td>University of Sheffield, Sheffield, UK</td>
<td>Ultrafast optics on quantum dots</td>
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<td>Pere Roca i Cabarrocas</td>
<td>Ecole Polytechnique, Palaiseau, France</td>
<td>Synthesis of silicon nanowires</td>
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<td>Lars Samuelson</td>
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<td>Nobu Sawaki</td>
<td>Nagoya University, Japan</td>
<td>Electronic structure of quantum dots</td>
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<td>Giorgio Sberveglieri</td>
<td>University of Brescia, Italy</td>
<td>Metal oxide chemical sensors</td>
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<td>Michelle Simmons</td>
<td>University of New South Wales, Australia</td>
<td>Single donor devices</td>
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<td>Paulo Sérgio Soares Guimarães</td>
<td>Federal University of Minas Gerais, Brazil</td>
<td>Theory of interband optical properties of quantum dots</td>
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<td>Tadeusz Suski</td>
<td>Polish Academy of Sciences, Warsaw, Poland</td>
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<td>Sasha Tartakovskii</td>
<td>University of Sheffield, Sheffield, UK</td>
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<td>Eric Tournie</td>
<td>University of Montpellier 2, Montpellier, France</td>
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<td>Sergio Ulloa</td>
<td>Ohio University, Ohio, USA</td>
<td>Spin polarization control via magnetic barriers and spin-orbit effects</td>
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<td>Selim Ünlü</td>
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<td>Karl Unterrainer</td>
<td>Technische Universität Wien, Austria</td>
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<td>Milan Vanecek</td>
<td>Fyzikalni ustav, Prague, Czech Republic</td>
<td>Optical characterization of diamond films</td>
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<td>Tim Veal</td>
<td>University of Warwick, Physics Dept., UK</td>
<td>X-ray photoelectron spectroscopy on III-V nanowires</td>
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<td>Naoki Yokoyama, Shozo Fujita</td>
<td>Fujitsu Laboratories Ltd, Atsugi, Japan</td>
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<td>Jinping Zhang</td>
<td>Suzhou Institute of Nano-Tech and Nano-Bionics, Chinese Academy of Sciences</td>
<td>Transmission electron microscopy on III-N nanowires</td>
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<td>Ning-Hua Zhu</td>
<td>Institute of Semiconductors, Chinese Academy of Sciences, Beijing</td>
<td>High-speed VCSELs, optical MIMOs</td>
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<td>Andreas Bausch</td>
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<td>Maximilian Fleischer</td>
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<td>Alois Friedberger</td>
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<td>Peter Hänggi</td>
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<td>Achim Hartschuh</td>
<td>Ludwig-Maximilians-Universität München</td>
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<td>Khaled Karrai</td>
<td>Attocube, Munich</td>
<td>Transmission experiments on quantum dots</td>
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<td>Klaus von Klitzing, Jürgen Smet, Werner Dietsche</td>
<td>MPI für Festkörperforschung Stuttgart</td>
<td>Coupled 2d Systems, quantum Hall systems</td>
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<td>Patrick Kölsch</td>
<td>Universität Heidelberg</td>
<td>SFG spectroscopy: diamond/water interfaces</td>
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<td>Erhard Kohn</td>
<td>Universität Ulm</td>
<td>Diamond biosensors</td>
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<tr>
<td>Jörg Peter Kotthaus</td>
<td>Ludwig-Maximilians-Universität München, CeNS, München</td>
<td>GaAs based hetero- and nanostructures, simulation of gated quantum dots</td>
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<td>Alois Krost</td>
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<td>Armin Lambrecht</td>
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<td>Karin Larsson</td>
<td>Uppsala University</td>
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<td>John Foord</td>
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<td>Ken Haenen</td>
<td>IMEC-IMOMEC</td>
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<td>U. Johann</td>
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<td>Jaromir Rehak</td>
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<td>Doris Steinmüller-Nethl</td>
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<td>Song Li</td>
<td>Ludwig-Maximilians-Universität München</td>
<td>CVD-growth of carbon nanostructure</td>
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<td>Paolo Lugli</td>
<td>Technische Universität München</td>
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<td>Peter Meissner</td>
<td>Technische Universität Darmstadt</td>
<td>Widely tunable MEMS-VCSEL</td>
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<td>Bruno K. Meyer</td>
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<td>Spektroskopie von ZnMgO</td>
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<td>Ulrich Schmid / Helmut Seidel</td>
<td>Universität des Saarlandes, Saarbrücken</td>
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<td>J. M. Schoening</td>
<td>Aachen University of Applied Sciences</td>
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<td>Matthias Schreck</td>
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<td>Diether Schuh</td>
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<td>Robert Tampé</td>
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<td>Marc Tornow</td>
<td>Universität Braunschweig</td>
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<td>Achim Trampert</td>
<td>Paul-Drude-Institut, Berlin</td>
<td>Transmission electron microscopy on GeMn</td>
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<td>Werner Wegscheider</td>
<td>Universität Regensburg</td>
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<td>Carsten Werner</td>
<td>Max Bergmann Center of Biomaterials, Dresden</td>
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<td>Artur Zrenner</td>
<td>Universität Paderborn</td>
<td>Single quantum dot photo diodes</td>
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4. Members of the Institute, Postdocs, and Guest Scientists

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Alexander Efros (Humboldt Senior Fellow)
Simon Frederick (Humboldt Fellow)
Thomas Grange (Humboldt Fellow)
Aki Michihiko, Fujitsu Laboratories Ltd, Atsugi, Japan
Vladimir Mitin (Humboldt Senior Fellow)
Hiroki Morishita, Keio University, Yokohama, Japan
Michael Shur (Humboldt Senior Fellow)
Jose M. Villas-Boas (Humboldt Fellow)
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Tobias Antesberger  Konrad Klein
Shamsul Arafı       Florian Klotz
Alexander Bachmann  Jelena Knezevic
Silvia Baldovino    Tillmann Kubis*
Barbara Baur        Arne Laucht
Christoph Bihler*   Bernhard Laumer
Stefan Birner       Robert Lechner*
Gregor Bracher      Simon Lud
Felix Buth          Markus Mangold
Jia Chen            Makiko Maruyama
Emily Clark         Kai Müller
Markus Banker       Michael Müller
Matthias Firnkes    Sabrina Niesar
Florian Furtmayr    Daniel Pedone
Wojciech Gajewski   Leonhard Prechtel
Tonko Garma         Andreas Reitinger
Christian Grasse    Johanna Rössler, geb. Simon*
Peter Greck         Jürgen Sailer
Tobias Gründl       Susanne Schäfer
Paul Hampel         Christoph Schindler
Andreas Hangauer    Sebastian Schöll
Moritz Hauf         Michael Scholz
Norman Hauke        Markus Schuster
Dominik Heiss*      Alexander Schwemer
Martin Heiß         Narayan Sircar
Martin Hermann*     Dance Spirkoska
Simon Hertenberger  Markus Stallhofer
Marco Höb           Andre Stegner
Felix Höhne         Lucia Steinke*
Felix Hofbauer*     Benedikt Stoib
Werner Hofmann*     Sebastian Strobel*
John Howgate        Jörg Teubert
Nebile Isik*        Augustinas Vizbaras
Christian Jäger     Kristijonas Vizbaras
Vase Jovanov        Thomas Wassner
Wolfgang Kaiser     Ruoshan Wei
Michael Kaniber*    Olaf Weidemann
Simone Kaniber      Andreas Wild
                    Ying Xiang*
                    Thomas Zabel
                    Ilaria Zardo
                    Vera Zon
Diploma Students (thesis completed*)
Matthias Anders                  Jinming Lu*
Marianne Auernhammer             Michael Maier*
Mario Baréiβ*                    Johannes Manz
Christian Böttner *               Johannes Mösl*
Matthias Brandstetter*           Kai Müller*
Carlo Diaz Alvarez*              Alexandra Münzer
Valentin Döring*                 Robin Nagel
Franz Eppe                       Helmut Nesswetter*
Nadine Erhard                    Christoph Pellinger
Najafi Faraz                     Günther Reithmaier
Emanuel Forster                  Daniel Rudolph*
Thomas Hager                     Rebecca Saive
Paul Hampel*                     Daniel Schnurbusch
Markus Hanl*                     Konrad Schönleber*
Benedikt Hauer                   Fabian Schuster
Norman Hauke*                    Max Seifert
Birgit Hausmann*                 Martin Soini
Martina Hofmann                  Christoph Stark*
Christian Huck                   Benedikt Stoib*
Malte Huck*                      Sebastian Thunich*
Eike Janocha*                    Marcel Törpe
Stefan Kapfinger                 Markus Trautwein*
Johannes Kierig                  Christoph Weiβ*
Marion Kraus*                    Ruoshan Wei*
Andreas Langer                   Andreas Wild*
Hagen Langhuth*                  Doris Wozniak
Stefan Lichtmannecker

Master Students
Abbas Mehtashami*                Xie Qiushi
Cate Morgan*                     Augustinas Vizbaras*
Swarnakamal Priyabadini*         Kristijonas Vizbaras*

Bachelor Students
Christopher Birkenstock*         Benedikt Mayer
Tobias Häberle*                  Kai Saller*
Christoph Kastl*                 Konrad Schraml*
Sarah Lindner*                   Ulrich Schwarzenberger*
5. **Doctoral, Diploma, Bachelor, and Master Theses**

**Doctoral Theses**

1. *Realization of electrically tunable single quantum dot nanocavities*  
   Felix Florian Georg Hofbauer (7.1.2009)

2. *Metal-insulator transition and superconductivity in heavily boron-doped diamond and related materials*  
   Philipp Achatz (15.1.2009)

3. *InP-based long-wavelength VCSELs and VCSEL arrays for high-speed optical communication*  
   Werner Hofmann (26.1.2009)

4. *Silicon nanocrystal films for electronic applications*  
   Robert W. Lechner (6.2.2009)

5. *Investigation of Landau level spin reversal in (110) oriented p-type GaAs quantum wells*  
   Nebile Isik (17.2.2009)

6. *Spatially resolved spectroscopy on semiconductor nanostructures*  
   Johanna Rössler, geb. Simon (26.2.2009)

7. *Optoelectronic and spin-related properties of semiconductor nanostructures in magnetic fields*  
   Till Andlauer (11.3.2009)

8. *Tunnel-Spektroskopie im Quanten-Hall-Regime*  
   Michael Huber (19.3.2009)

9. *Novel photonic biosensing based on silicon nanostructures*  
   Dominic F. Dorfner (1.4.2009)

10. *Nanoscale contacts to organic molecules based on layered semiconductor substrates*  
    Sebastian Strobel (3.4.2009)

11. *Magnetic semiconductors*  
    Christoph Bihler (7.4.2009)

12. *Epitaktische AlN-Schichten auf Saphir und Diamant*  
    Martin Hermann (27.4.2009)

13. *Magnetotransport of coupled quantum Hall edges in a bent quantum well*  
    Lucia Steinke (19.6.2009)
14. Semiconductor nanowires and templates for electronic applications
Ying Xiang (23.6.2009)

15. Non-classical light generation in photonic crystal nanostructures
Michael Kaniber (7.7.2009)

16. Spin storage in quantum dot ensembles and single quantum dots
Dominik Heiss (30.9.2009)

17. Quantum transport in semiconductor nanostructures
Tillmann Christoph Kubis (29.10.2009)

Diploma Theses

1. sEDMR – A novel spatially resolved magnetic resonance technique
Markus Michael Trautwein (1/2009)

2. Electrically tunable single quantum dot nanocavities in the strong coupling regime
Norman Hauke (2/2009)

3. Nanoscale imaging of surfaces by electrochemical scanning tunneling microscope
Philipp Sebastian Koch (2/2009)

4. Epitaxie und Charakterisierung von InAs/GaInAs Quantenpunkten für eine
Einzelphotonenquelle bei 1.55 µm
Marion Kraus (2/2009)

5. Kontaktieren von organischen und anorganischen Nanosystemen mittels
Kohlenstoffnanoröhrchen
Matthias Brandstetter (4/2009)

6. Materials for electron g-factor switching in Si/SiGe heterostructures
Andreas Wild (4/2009)

7. Chemische Oberflächenmodifizierung von Siliciumcarbid
Carlos Albert Díaz Álvarez (5/2009)

8. Optical and electrical properties of single self-assembled quantum dots in lateral
electric fields
Malte Huck (6/2009)

9. Photonic crystal nanocavities as unique objects
Kai Müller (6/2009)
10. Komplexe nano-plasmonische Hybridmaterialien als einzigartige Objekte in der physikalischen Kryptographie
Hagen Langhuth (7/2009)

11. Thermoelektrische Eigenschaften dünner Schichten aus Silizium- und Germanium-Nanopartikeln
Konrad Schönleber (7/2009)

12. Optoelektronische Eigenschaften von GaAs-Nanodrähten
Sebastian Thunich (7/2009)

13. Diamond nanowires for quantum information processing applications
Birgit Hausmann (10/2009)

14. Nanocontacts for spatially resolved electrically detected magnetic resonance
Benedikt Stoib (10/2009)

15. Anorganische Halbleiter – P3HT Heteroübergänge für Solarzellen
Helmut Nesswetter (10/2009)

16. Optoelektronische Eigenschaften von zweidimensionalen Gold-Nanopartikel-Superstrukturen
Christoph Weiß (10/2009)

17. Random pn-junctions for physical cryptography
Michael Andreas Algasinger (11/2009)

18. Spin properties of isotopically controlled phosphorus-doped silicon epilayers
Jinming Lu (11/2009)

19. InAs(Sb)-Quantenpunkte auf InP-Substrat für eine Einzelphotonenquelle bei 1,55µm
Mario Bareiß (11/2009)

20. Development and realization of waveguide concepts for InP based Terahertz quantum cascade lasers
Gregor Bracher (11/2009)

21. Development and characterization of active regions for THz quantum cascade lasers based on InP
Valentin Döring (11/2009)

22. Isolation von DNA-Goldnanopartikel-Konjugaten für Energietransferstudien und plasmonische Anwendungen
Christian Böttner (12/2009)
23. *Dynamics of surface-tethered protein-DNA conjugates in AC electric fields*
   Paul Hampel (12/2009)

24. *Optical spin storage and readout in a single self-assembled quantum dot*
   Daniel Rudolph (12/2009)

25. *Fabrication and functionalization of metalized nanopores*
   Ruoshan Wei (12/2009)

**Bachelor Theses**

1. *A micron-sized nanoporous multifunction sensing device*
   Christopher Birkenstock (2/2009)

2. *Gating of two-dimensional electron systems in Si/SiGe*
   Christoph Kastl (5/2009)

3. *Photocurrent dynamics of silicon wafers*
   Ulrich Schwarzenberger (7/2009)

4. *Optical and structural investigations of ZnMgO thin films*
   Sarah Lindner (8/2009)

5. *Photocurrent spectroscopy of CdTe single crystals*
   Konrad Schraml (8/2009)

6. *Alkylated GaN surfaces*
   Tobias Haeberle (8/2009)

7. *Investigation of recombination mechanisms in Gallium antimonide based lasers*
   Kai Saller (10/2009)

**Master Theses**

1. *Development of a terahertz laser source based on intracavity difference frequency generation in quantum cascade lasers*
   Augustinas Vizbaras (1/2009)

2. *Development of active regions for GaSb based VCSELs, emitting at wavelengths above 2.3 µm*
   Kristijonas Vizbaras (1/2009)
3. *Low resistive ohmic contacts for GaSb-based VCSELs in the mid-infrared range*  
   Swarnakamal Priyabadini (5/2009)

4. *Preparation and properties of platinum nanoparticles on wide bandgap semiconductor surfaces*  
   Yizhen Wang (5/2009)

6. *Characterization of silicon nanowires grown using indium or gallium as VLS catalysts*  
   Cate Morgan (5/2009)

7. *Development of biosensors based on electrically switchable PNA/DNA interfaces*  
   Kiran Girdhar (6/2009)

8. *Investigations of off-resonant dot-cavity coupling in photonic crystal nanocavities*  
   Abbas Mohtashami (11/2009)
6. Publications

*Ab initio calculation of the dispersion interaction between a polyaromatic molecule and a noble metal substrate: PTCDA on Ag(110)*
A. Abbasi and R. Scholz

*Metal-insulator transition and superconductivity in highly boron-doped nanocrystalline diamond films*

*InP-based long-wavelength VCSELs and VCSEL arrays*
M.-C. Amann and W. Hofmann

*Low-temperature transport in highly boron-doped nanocrystalline diamond*
P. Achatz, W. Gajewski, E. Bustarret, C. Marcenat, R. Piquerel, C. Chapelier, T. Dubouchet, O. A. Williams, K. Haenen, J. A. Garrido, and M. Stutzmann

*Comparison of the magnetic properties of GeMn thin films through Mn L-edge x-ray absorption*
S. Ahlers, P. R. Stone, N. Sircar, E. Arenholz, O. D. Dubon, and D. Bougeard

*Electrically controllable g tensors in quantum dot molecules*
T. Andlauer and P. Vogl

*Full-band envelope function approach for type-II broken-gap superlattices*
T. Andlauer and P. Vogl

*Self-consistent electronic structure method for broken-gap superlattices*
T. Andlauer, T. Zibold, and P. Vogl
Measuring the charge and spin states of electrons on individual dopant atoms in silicon

Continuous-wave electrically-pumped GaSb-based VCSELs at ~ 2.6 µm operating up to 50 °C
S. Arafin, A. Bachmann, K. Kashani-Shirazi, and M.-C. Amann
The 22nd Annual Meeting of the IEEE Photonics Society, Belek-Antalya, Turkey (2009)

Continuous-wave single-mode electrically-pumped GaSb-based VCSELs at 2.5 µm
S. Arafin, A. Bachmann, K. Kashani-Shirazi, and M.-C. Amann
Conference on Lasers and Electro-Optics (CLEO)-Pacific Rim, Shanghai, China (2009)

Electrically-pumped continuous-wave vertical-cavity surface-emitting lasers at ~ 2.6 µm
S. Arafin, A. Bachmann, K. Kashani-Shirazi, and M.-C. Amann
Appl. Phys. Lett. 95, 131120, 1-3 (2009)

Low-resistive ohmic contacts to n- InAs0.91Sb0.09 for GaSb-based VCSELs in the mid-infrared range
S. Arafin, A. Bachmann, K. Kashani-Shirazi, S. Priyabadini, and M.-C. Amann
Semiconductor and Integrated Opto-Electronics Conference (SIOE), Cardiff, United Kingdom (2009)

Low-resistive sulphur-treated ohmic contacts to n-type InAsSb
S. Arafin, A. Bachmann, K.Kashani-Shirazi, S. Priyabadini, and M.-C. Amann
IET Optoelectron. Selected papers inspired by the Semiconductor and Integrated Optoelectronics Conference (SIOE), 3, 6, 259-263 (2009)

Triple-twin domains in Mg doped GaN wurtzite nanowires: structural and electronic properties of this zinc-blende-like stacking
Nanotechnology 20, 145704 (2009)

GaSb-based VCSEL with buried tunnel junction for emission around 2.3 µm
A. Bachmann, K. Kashani-Shirazi, S. Arafin, and M.-C. Amann
Long wavelength electrically pumped GaSb-based buried tunnel junction VCSELs
A. Bachmann, S. Arafin, K. Kashani-Shirazi, and M.-C. Amann
International Conference on Narrow Gap Semiconductors and Systems (NGSS), Sendai, Japan (2009)

Long wavelength electrically pumped GaSb-based buried tunnel junction VCSELs
A. Bachmann, S. Arafin, K. Kashani-Shirazi, and M.-C. Amann
Physica Procedia 3, Issue 2, 1155-1159 (2009)

Single-mode electrically pumped GaSb-based VCSELs emitting continuous-wave at 2.4 and 2.6 µm
A. Bachmann, S. Arafin, and K. Kashani-Shirazi

Antimon assisted growth of InAs quantum dots
M. Bareiss, R. Enzmann, R. Meyer, G. Böhm, J. Finley, and M.-C. Amann
Deutscher MBE-Workshop, Bochum, Germany (2009)

Quantum dots on AlGaInAs lattice matched to InP emitting at 1.55 µm
M. Bareiss, R. Enzmann, G. Böhm, J. Finley, R. Meyer, and M.-C. Amann
NIM/CeNS/SFB 486 Winterschool, Nanosystems and Sensors, St. Christoph, Austria (2009)

Ultrathin GaN/AlN/GaN solution-gate field effect transistor with enhanced resolution at low source-gate voltage
A. Bengoechea Encabo, J. Howgate, M. Stutzmann, M. Eickhoff, and M. A. Sánchez-Garcia

Manganese-hydrogen complexes in Ga1-xMn1-xN

Spin-wave resonances and surface spin pinning in GaMnAs thin films
C. Bihler, W. Schoch, W. Limmer, S. T. B. Goennenwein, and M. S. Brandt

Ballistic quantum transport using the contact block reduction (CBR) method
S. Birner, C. Schindler, P. Greck, M. Sabathil, and P. Vogl

Al(In)As-(Ga)InAs strain-compensated active regions for injectorless quantum cascade lasers
G. Böhm, S. Katz, R. Meyer, and M.-C. Amann
**Ge1-x Mnx clusters: Central structural and magnetic building blocks of nanoscale wire-like self-assembly in a magnetic semiconductor**  

**Donors for quantum information processing**  
M. S. Brandt  
in: Spintronics – From GMR to Quantum Information, Lecture Notes of the 40th Spring School of the Institute of Solid-State Research, Forschungszentrum Jülich, hrsg. von S. Blügel, D. Bürgler, M. Morgenstern, C. M. Schneider, and R. Waser (2009), S. E3.1

**CO and CH₄ sensing with single mode 2.3 μm GaSb-based VCSEL**  
J. Chen, A. Hangauer, A. Bachmann, T. Lim, K. Kashani-Shirazi, R. Strzoda, and M.-C. Amann  

**VCSEL-based oxygen sensor for combustion optimization in gas/oil furnaces**  
J. Chen, A. Hangauer, R. Strzoda, M. Fleischer, and M.-C. Amann  

**Tuning magnetic properties of magnetoelectric BiFeO₃-NiFe₂O₄ nanostructures**  
S. P. Crane, C. Bihler, M. S. Brandt, S. T. B. Goennenwein, M. Gajek, and R. Ramesh  

**Diamond transistor array for extracellular recording from electrogenic cells**  

**Light-induced charge transfer in hybrid composites of organic semiconductors and silicon nanocrystals**  

**Outcoupling of light generated in a monolithic silicon photonic crystal nanocavity through a lateral waveguide**  
D. Dorfner, S. Iwamoto, M. Nomura, S. Nakayama, J. Finley, G. Abstreiter, and Y. Arakawa  

**Photonic crystal nanostructures for optical biosensing applications**  
D. Dorfner, T. Zabel, T. Hürlimann, N. Hauke, L. Frandsen, U. Rant, G. Abstreiter, and J. Finley  
Biosensors and Bioelectronics 24, 3688 (2009)
Ultrathin GaN/AlN/GaN solution-gate field effect transistor with enhanced resolution at low source-gate voltage
A. Encabo, J. Howgate, M. Stutzmann, M. Eickhoff, and M. A. Sánchez-García
Sensors and Actuators B: Chemical 142, 1, 304-307 (2009)

Concept of a single photon source based on InP for emission at 1.55 µm
R. Enzmann, C. Jendrysik, M. Bareiss, C. Grasse, G. Böhm, R. Meyer, J. Finley, and M.-C. Amann
International Conference on Nanomaterials and Nanosystems (NanoMats 2009)
Istanbul, Turkey (2009)

Growth of low-density quantum dots emitting at telecommunication wavelengths
R. Enzmann, M. Bareiss, M. Kraus, G. Böhm, J. Finley, R. Meyer, and M.-C. Amann
Semiconductor and Integrated Opto-Electronics Conference (SIOE), Cardiff, United Kingdom (2009)

InAs quantum dots on AlGaInAs emitting in the optical C-band at 1.55 µm
R. Enzmann, M. Kraus, M. Bareiss, C. Seidel, D. Baierl, G. Böhm, R. Meyer, J. Finley, and M.-C. Amann
International Conference on Indium Phosphide and Related Materials (IPRM),
Newport Beach, CA, USA (2009)

Concept of a single photon source based on InP for emission at 1.55 µm
R. Enzmann, M. Bareiss, G. Böhm, R. Meyer, J. Finley, and M.-C. Amann
NIM/CeNS/SFB 486 Winterschool, Nanosystems and Sensors, St. Christoph, Austria
(2009)

Optical glucose monitoring using vertical cavity surface emitting lasers (VCSELs)
S. T. Fard, W. Hofmann, E. Kwok, M.-C. Amann, and L. Chrostowski
Optics and Photonics Symposium, Bio-Sensing II, Novel and Bio-inspired Sensors,
San Jose, CA, USA, 7397, pp. 739704-739711 (2009)

Investigation of a contacting scheme for self-assembled cleaved edge overgrown InAs nanowires and quantum dot arrays
M. Fehr, E. Uccelli, S. Dasgupta, M. Bichler, L. Steinke, G. Abstreiter, M. Grayson, and A. Fontcuberta i Morral
phys. stat. sol. (a) 206, 1620 (2009)

Photoresponse and morphology of pentacene thin films modified by oxidized and reduced diamond surfaces
W. Gajewski, M. Huth, F. Buth, B. Nickel, M. Stutzmann, and J. A. Garrido

Electronic and optical properties of boron-doped nanocrystalline diamond films
W. Gajewski, O. A. Williams, P. Achatz, K. Haenen, E. Bustarret, M. Stutzmann, and J. A. Garrido
Crystallochromy of perylene pigments: Interference between Frenkel excitons and charge-transfer states
L. Gisslen and R. Scholz

Magnetic anisotropy in (Ga,Mn)As: Influence of epitaxial strain and hole concentration

Decoherence in quantum dots due to real and virtual transitions: A nonperturbative calculation
T. Grange

Theory of relaxation and decoherence of intersublevel transitions in semiconductor quantum dots
T. Grange, R. Ferreira, and G. Bastard

Planarization of overgrown tunnel junctions for InP-based VCSEL by MOVPE
C. Grasse, M. Müller, G. Böhm, R. Enzmann, Y. Xu, M. Görblich, R. Meyer, M. Ortsiefer, and M.-C. Amann
European Workshop on Metalorganic Vapor Phase Epitaxy (EWMOVPE), Ulm, Germany (2009)

Anomalous modulation characteristics of optical injection-locked VCSELs
P. Guo, W. Yang, D. Parekh, W. Hofmann, M.-C. Amann, and C. J. Chang-Hasnain
Asia Communications and Photonics Conference and Exhibition (ACP), Shanghai, China (2009)

Influence of hydrogen on nanocrystalline diamond surfaces investigated with HREELS and XPS
phys. stat. sol. (a) 206, 9, 2022-2027 (2009)

Fire detection with a compact 2.3 µm VCSEL-based carbon monoxide sensor
A. Hangauer, J. Chen, R. Strzoda, M. Fleischer, and M.-C. Amann

Temperature and doping dependence of electron spin relaxation in GaSb
C. Hautmann, F. Jaworeck, K. Kashani-Shirazi, M. C. Amann, and M. Betz
Long range epitaxial growth of prismatic heterostructures on the facets of catalyst-free GaAs nanowires

Selective optical charge generation, storage, and readout in a single self-assembled quantum dot
D. Heiss, V. Jovanov, M. Caesar, M. Bichler, G. Abstreiter, and J. Finley

Catalyst-free nanowires with axial InxGa1-xAs/GaAs heterostructures’
Nanotechnology 20, 075603 (2009)

Electrochemical impedance spectroscopy of oxidized and hydrogen-terminated nitrogen-induced conductive ultrananocrystalline diamond

Photoconductive gain in semiconductor quantum wires
arxive: 0810.0970 (2009)

1.55 µm InP-based VCSEL with enhanced modulation bandwidths > 10 GHz up to 85 °C
W. Hofmann, M. Müller, G. Böhm, M. Ortsiefer, and M.-C. Amann
Optical Fiber Communication Conference and Exposition (OFC) and The National Fiber Optic Engineers Conference (NFOEC), San Diego, USA (2009)

1.55 µm VCSEL with enhanced modulation bandwidth and temperature range
W. Hofmann, M. Müller, G. Böhm, M. Ortsiefer, and M.-C. Amann

22-Gb/s long wavelength VCSELs
W. Hofmann, M. Müller, A. Nadtochiy, C. Meltzer, A. Mutig, G. Böhm, J. Rosskopf, D. Bimberg, M.-C. Amann, and C. J. Chang-Hasnain

Long-wavelength BTJ-VCSELs with improved modulation bandwidth and temperature range
W. Hofmann, M. Müller, G. Böhm, J. Rosskopf, and M.-C. Amann
**Long-wavelength high-contrast grating vertical-cavity surface-emitting laser**
W. Hofmann, C. Chase, M. Müller, Y. Rao, G. Böhm, M.-C. Amann, and C. J. Chang-Hasnain

**1.55-μm VCSEL modulation performance with dispersion-compensating fibers**
W. Hofmann, L. Grüner-Nielsen, E. Rönneberg, G. Böhm, M. Ortsiefer, and M.-C. Amann

**Long-wavelength VCSEL arrays with partly coherent emission**
W. Hofmann, M. Görblich, G. Böhm, M. Ortsiefer, and M.-C. Amann
Conference on Lasers and Electro-Optics CLEO/QELS, Baltimore, USA, CMSS-VCSELs II, CMSS3 (2009)

**1.55-μm VCSEL arrays for optical multiple-input multiple-output (MIMO)**
W. Hofmann, N. H. Zhu, M. Görblich, L. Xie, G. Böhm, M. Ortsiefer, and M.-C. Amann
Conference on Lasers and Electro-Optics CLEO/QELS, Baltimore, USA, JTuD17 (2009)

**Phonon-assisted transitions from quantum dot excitons to cavity photons**
U. Hohenester, A. Laucht, M. Kaniber, N. Hauke, A. Neumann, A. Mohtashami, M. Seliger, M. Bichler, and J. Finley

**Mid-infrared optical response of heavily doped GaSb:Te**
J. Humlicek, K. Navrtil, E. Hulicius, J. Vyskocil, T. Simecek, K. Kashani-Shirazi, and M.-C. Amann
International Conference on Mid-Infrared Optoelectronics, Freiburg, Germany, p. 164-5 (2009)

**Gain peak-cavity mode alignment optimisation in buried tunnel junction mid-infrared GaSb vertical cavity surface emitting lasers using hydrostatic pressure**
IET Optoelectron. Selected Papers inspired by the Semiconductor and Integrated Optoelectronics Conference (SIOE) **3**, 6, 305-309 (2009)

**Efficient and selective cavity-resonant excitation for single photon generation**
M. Kaniber, A. Neumann, A. Laucht, M. Huck, M. Bichler, M.-C. Amann, and J. Finley

**MBE growth of active regions for electrically-pumped, cw-operating GaSb-based VCSELs**
K. Kashani-Shirazi, A. Bachmann, G. Böhm, S. Ziegler, and M.-C. Amann
Low-threshold strained quantum-well GaSb-based lasers emitting in the 2.5 to 2.7 µm wavelength range  
K. Kashani-Shirazi, K. Vizbaras, A. Bachmann, S. Arafin, and M.-C. Amann  

Optimized MBE growth technique for GaSb-based edge emitters at 2.7 µm  
K. Kashani-Shirazi, A. Bachmann, S. Arafin, K. Vizbaras, S. Ziegler, and M.-C. Amann  
EURO-MBE, Zacopane, Poland (2009)

Ultra-low threshold GaSb-based laser diodes at 2.65 µm  
K. Kashani-Shirazi, A. Bachmann, S. Arafin, K. Vizbaras, and M.-C. Amann  

High power injectorless quantum cascade laser structure in the 6.0 µm wavelength range  
S. Katz, G. Böhm, and M.-C. Amann  

High-performance injectorless quantum cascade laser emitting below 6 µm  
S. Katz, A. Vizbaras, G. Böhm, and M.-C. Amann  

Injectorless quantum cascade laser operating in continuous wave above room temperature  
S. Katz, A. Vizbaras, G. Böhm, and M.-C. Amann  

Magnetic field dependence of dielectric polarization echoes in KCl doped with Li  
F. Klotz, A. Fleischmann, and C. Enss  
J. Phys. CS 150, 042028 (2009)

Ordered Si nanoaperture arrays for the measurement of ion currents across lipid membranes  
S. Koynov, M. S. Brandt, and M. Stutzmann  
Appl. Phys. Lett. 95, 023112 (2009)

How periodic are terahertz quantum cascade lasers?  
T. Kubis and P. Vogl  

Predictive quantum theory of current and optical gain in quantum cascade lasers  
T. Kubis and P. Vogl  
Laser Physics 19, 762 (2009)

Predictive quantum theory of current and optical emission in quantum cascade lasers  
T. Kubis and P. Vogl  
Theory of non-equilibrium quantum transport and energy dissipation in terahertz quantum cascade lasers
T. Kubis, C. Yeh, P. Vogl, A. Benz, G. Fasching, and C. Deutsch

Gate control of low-temperature spin dynamics in two-dimensional hole systems
M. Kugler, T. Andlauer, T. Korn, A. Wagner, S. Fehringer, R. Schulz, M. Kubová, C. Gerl, D. Schuh, W. Wegscheider, P. Vogl, and C. Schüller

Optical spectra obtained from amorphous films of rubrene: Evidence for predominance of twisted isomer
M. Kytka, L. Gisslen, A. Gerlach, U. Heinemeyer, J. Kovac, R. Scholz, and F. Schreiber

Thermal radiation of two-dimensional Bose-Einstein gas of surface plasmons
A. N. Latyshev, D. A. Minakov, O. V. Ovchinnikov, V. A. Buslov, O. G. Vikin, V. Zon, and B. A. Zon

Dephasing of exciton polaritons in photoexcited InGaAs quantum dots in GaAs nanocavities
A. Laucht, N. Hauke, J. M. Villas-Boas, F. Hofbauer, G. Boehm, M. Kaniber, and J. Finley

Electrical control of spontaneous emission and strong coupling for a single quantum dot
A. Laucht, F. Hofbauer, N. Hauke, J. Angele, S. Stobbe, M. Kaniber, G. Boehm, P. Lodahl, M. C. Amann, and J. Finley

Surface plasmon enhanced photoconductance of gold nanoparticle arrays with incorporated alkane linkers
M. Mangold, C. Weiss, M. Calame, and A. Holleitner

Buried tunnel junction mid-infrared GaSb VCSELs investigated using hydrostatic pressure
Semiconductor and Integrated Opto-Electronics Conference (SIOE), Cardiff, United Kingdom (2009)

Carrier transport in THz quantum cascade lasers: Are Green's functions necessary?
A. Mátyás, T. Kubis, P. Lugli, and C. Jirauschek
1.55 µm InP-based short-cavity-VCSEL with enhanced modulation-bandwidth of 15 GHz
M. Müller, W. Hofmann, G. Böhm, J. Rosskopf, E. Rönneberg, M. Ortsiefer, and M.-C. Amann
European Conference on Optical Communications (ECOC), Vienna, Austria, Paper 8.2.1 pp. 1-2 (2009)

Short-cavity long-wavelength VCSELs with modulation bandwidths in excess of 15 GHz
M. Müller, W. Hofmann, G. Böhm, and M.-C. Amann

InP-based VCSEL with novel short-cavity design for enhanced modulation bandwidth
M. Müller, W. Hofmann, and M.-C. Amann
International Nano-Optoelectronics Workshop (i-NOW), Stockholm, Sweden / Berlin, Germany, 199-200 (2009)

Light-induced charge transfer in hybrid composites of organic semiconductors and silicon nanocrystals
S. Niesar, R. Dietmueller, H. Nesswetter, H. Wiggers, and M. Stutzmann
phys. stat. sol. (a) 206, 2775 (2009)

90-km single mode fiber transmission of 10-Gb/s multimode VCSELs under optical injection locking
D. Parekh, B. Zhang, X. Zhao, Y. Yue, W. Hofmann, M.-C. Amann, A. E. Willner, and C. J. Chang-Hasnain
OFC/NFOEC, San Diego, USA (2009)

Isolator-less optically injection-locked 1.55-µm VCSELs for upstream transmitters in WDM-Pons
D. Parekh, W. Yang, W. Hofmann, M.-C. Amann, and C. J. Chang-Hasnain
OFC/NFOEC, San Diego, USA (2009)

Dielectric screening versus quantum confinement of phosphorus donors in silicon nanocrystals investigated by magnetic resonance
R. N. Pereira, A. R. Stegner, T. Andlauer, K. Klein, H. Wiggers, M. S. Brandt, and M. Stutzmann

Investigation of interband optical transitions by near-resonant magneto-photoluminescence in InAs/GaAs quantum dots
V. Preisler, T. Grange, R. Ferreira, et al.
Low-resistive thermally stable metal-semiconductor contacts on n-GaSb using n-InAsSb contact layer
S. Priyabadini, S. Arafín, A. Bachmann, K. Kashani-Shirazi, and M.-C. Amann
International Conference on Frontiers of Physics (IFCP), Kathmandu, Nepal (2009)

Detection and size analysis of proteins with switchable DNA layers

Analysis of polarization-dependent photoreflectance studies for c-plane GaN films grown on a-plane sapphire
M. Röppische, R. Goldhahn, C. Buchheim, F. Furtmayr, T. Wassner, M. Eickhoff, C. Cobet, and N. Esser
phys. stat. sol. (a) 206, 773 (2009)

Metal/organic interface formation studied in situ by resonant Raman spectroscopy
G. Salvan, B. A. Paez, D. R. T. Zahn, L. Gisslen, and R. Scholz

Prediction of giant intrinsic spin-Hall effect in strained p-GaAs quantum wells
A. Schindler and P. Vogl

Asymmetry between absorption and photoluminescence line shapes of TPD: Spectroscopic fingerprint of the twisted biphenyl core

Potentiometry on pentacene OFETs: Charge carrier mobilities and injection barriers in bottom and top contact configurations
R. Scholz, D. Lehmann, A. D. Müller, F. Müller, and D. R. T. Zahn

Gallium nitride electrodes for membrane based electrochemical biosensors

Cascaded exciton emission of an individual strain-induced quantum dot
F. J. R. Schülein, A. Laucht, J. Riikonen, M. Mattila, M. Sopanen, H. Lipsanen, J. Finley, A. Wixforth, and H. Krenner
Appl. Phys. Lett. 95, 083122 (2009)
Conjugated 12 nm long oligomers as molecular wires in nanoelectronics
J. Mat. Chem. 19, 3899 (2009)

Structural and optical properties of high quality zinc-blende/wurtzite GaAs nanowire heterostructures

Doping efficiency in freestanding silicon nanocrystals from the gas phase: Phosphorus incorporation and defect-induced compensation
B. R. Stegner, R. N. Pereira, R. Lechner, K. Klein, H. Wiggers, M. Stutzmann, and M. S. Brandt
Phys. Rev. B 80, 165326 (2009)

Planar nanogap electrodes by direct nanotransfer printing
S. Strobel, S. Harrer, G. Scarpa, G. Abstreiter, P. Lugli, and M. Tornow
Small 5, 579 (2009)

Photocurrent and photoconductance properties of a GaAs nanowire
S. Thunich, L. Prechtel, D. Spirkoska, G. Abstreiter, A. Fontcuberta i Morral, and A. Holleitner
Appl. Phys. Lett. 95, 083111 (2009)

Highly temperature-stable, long wavelength short-injector quantum cascade laser
A. Vizbaras, S. Katz, G. Böhm, and M.-C. Amann
International Conference on Intersubband Transitions in Quantum Wells (ITQW), Montreal, Canada (2009)

Short-injector quantum cascade laser emitting at 8 µm wavelength with high slope efficiency
A. Vizbaras, S. Katz, G. Böhm, and M.-C. Amann
IEEE Photonics Technology Letters 21, 1384 (2009)

Simultaneous two-level lasing in GaInAsSb/GaSb strained quantum-well lasers
K. Vizbaras, K. Kashani-Shirazi, and M.-C. Amann
Appl. Phys. Lett. 95, 071107 (2009)

Density enhanced diffusion of dipolar excitons within a one-dimensional channel
X. Vögele, D. Schuh, W. Wegscheider, J. Kotthaus, and A. Holleitner
Optical properties and structural characteristics of ZnMgO grown by plasma assisted molecular beam epitaxy
T. Wassner, B. Laumer, S. Maier, A. Laufer, B. Meyer, M. Stutzmann, and M. Eickhoff

GaN quantum dots as optical transducers for chemical sensors
O. Weidemann, P. K. Kandaswamy, E. Monroy, G. Jegert, M. Stutzmann, and M. Eickhoff

Voltage controlled inversion of magnetic anisotropy in a ferromagnetic thin film at room temperature
M. Weiler, A. Brandlmaier, S. Geprägs, M. Althammer, M. Opel, C. Bihler, H. Huebl, M. S. Brandt, R Gross und S. T. B. Goennenwein

Physical origin of data pattern inversion in optical injection-locked VCSELs
W. Yang, P. Guo, D. Parekh, W. Hofmann, M.-C. Amann, and C. J. Chang-Hasnain
Frontiers in Optics (FiO) 2009/Laser Science (LS) XXV Conference, San Jose, USA, (2009)

Nonequilibrium Green's function calculation for four-level scheme terahertz quantum cascade lasers
H. Yasuda, T. Kubis, P. Vogl, N. Sekine, I. Hosako, and K. Hirakawa

10-Gb/s colorless remodulation of signal from 1550nm vertical cavity surface emitting laser array in WDM Pon
L. Xu, H. K. Tsang, W. Hofmann, and M.-C. Amann
CLEO/Europe-EQEC Conference, Munich, Germany, Topic C1 (2009)

1.55-µm VCSEL transmission performance up to 20 Gb/s for access networks
L. Xu, W. Hofmann, H. K. Tsang, R. V. Penty, I. H. White, and M.-C. Amann
Optoelectronics and Communications Conference (OECC/IOOC), Hongkong, China, (2009)

Theory of nanocluster size distributions from ion beam synthesis
Phys. Rev. Lett. 102, 146101 (2009)

Size-distribution evolution of ion-beam-synthesized nanocrystals in silica
Phys. Rev. B 80, 134121 (2009)
Raman spectroscopy of wurtzite and zinc-blende GaAs nanowires: Polarization dependence, selection rules, and strain effects

Long lifetimes of quantum-dot intersublevel transitions in the terahertz range
E. A. Zibik, T. Grange, B. A. Carpenter, et al.

Reconfigurable multifunctional operation using optical injection-locked vertical-cavity surface-emitting lasers
B. Zhang, X. Zhao, D. Parekh, Y. Yue, W. Hofmann, M.-C. Amann, C. J. Chang-Hasnain, and A. E. Willner

Multifunctional and reconfigurable 10-GHz operation of an optical injection-locked VCSEL
B. Zhang, X. Zhao, D. Parekh, Y. Yue, W. Hofmann, M.-C. Amann, C. J. Chang-Hasnain, and A. E. Willner

Greatly increased fiber transmission distance with an optically injection-locked vertical-cavity surface-emitting laser
X. Zhao, D. Parekh, B. Zhang, L. Christen, A. E. Willner, W. Hofmann, M.-C. Amann, F. Koyama, and C. J. Chang-Hasnain
Opt. Express 17, 13785 (2009)

Coherent potential approximation for spatially correlated disorder
R. Zimmermann and C. Schindler
7. Invited Talks

Gerhard Abstreiter

1. *Einführung in die Nanowissenschaften*
   Paradigmenwechsel durch Nanowissenschaften, Tutzing, Germany (30.1.2009)

2. *GaAs based Nanostructures*
   University of Tokyo, Japan (13.2.2009)

   Lehrerfortbildung, Lüscher Seminar, Zwiesel, Germany (28.3.2009)

4. *Single charge and spin storage and readout in self-assembled InGaAs quantum dots*
   Kolloquium, University of California, Santa Barbara, USA (17.4.2009)

5. *Einführung in die Nanowissenschaften*
   Katherinen-Gymnasium Ingolstadt, Germany (29.4.2009)

6. *Semiconductor Quantum Dots – Optical Spectroscopy*
   QIP 2009, SFB 631, Herrsching, Germany (30.6.2009)

7. *Recent advances in GaAs based nanostructures with special emphasis on charge and spin storage and readout in quantum dots*
   Seminar, Fujitsu Lab., Atsugi, Japan (6.7.2009)

8. *Advanced GaAs based nanostructures*
   University of Tokyo, Komaba Campus, Japan (7.7.2009)

9. *Electronic states in semiconductor nanostructures*
   Lecture at University of Tokyo, Hongo Campus, Japan (13.7.2009)

10. *Nanoscience at Walter Schottky Institut of TU Munich: From semiconductor quantum dots and wires to biosensors*
    Toyota Lab., Nagoya, Japan (14.7.2009)

11. *Recent advances in GaAs based nanostructures with special emphasis on quantum wires and quantum dots*
    University of Kyoto, Japan (15.7.2009)

12. *Single charge and spin storage and readout in self-assembled InGaAs quantum dots*
    University of Tokyo, Hongo Campus, Japan (16.7.2009)
13. *Nanosystems for biosensing*
   NTT Basic Research Lab., Atsugi, Japan (27.7.2009)

14. *GaAs based nanostructures*
   NTT Basic Research Lab., Atsugi, Japan (27.7.2009)

15. *Single charge and spin storage and readout in self-assembled InGaAs quantum dots*
   Tohoku University, Sendai, Japan (28.7.2009)

16. *MBE grown GaAs based hetero-nanowires*
   Tohoku University, Sendai, Japan (29.7.2009)

17. *GaAs based nanostructures*
   Kejo University, Yokohama, Japan (30.7.2009)

18. *Nanoscience: From semiconductor quantum dots and wires to biosensors*
   Symposium on Nanoscience, Wroclaw, Poland (5.10.2009)

19. *Perpendicular versus parallel, research on tunneling and 2d-systems in the 1970th*
   Symposium on 50th Anniversary of Esaki Diode, Tokyo, Japan (18.11.2009)

20. *Electro-optic control of fundamental excitations in a self-assembled quantum dots*
    NanoQuine Symposium Tokyo, Japan (19.11.2009)

21. *Quanteneffekte in Nanometer-Halbleiterbauelementen*
    Bayerische Akademie der Wissenschaften, Munich, Germany (11.12.2009)

Markus-Christian Amann

1. *VCSEL structures and applications*
   IEEE Topical Meeting Innsbruck, Austria (13.1.2009)

2. *IR VCSELs*
   Infrarot Kolloquium Freiburg, Germany (17.2.2009)

3. *Single-mode and tunable VCSELs for communications and sensing*
   Technical University Eindhoven, The Netherlands (13.3.09)

4. *49 Jahre Laser – nun, wohin?*
   Universität Karlsruhe, Germany (22.7.2009)

5. *Tunable near- and mid-infrared VCSELs for sensing and spectroscopy*
Stefan Birner

1. *nextnano – a versatile tool for the simulation of nanostructures*

Dominique Bougeard

1. *Germanium based magnetic semiconductors*
   15th International Winterschool on New Developments in Solid State Physics, Bad Hofgastein, Austria (20.2.2008)

Martin S. Brandt

1. *The kinetics of light-induced degradation*
   International Workshop on Staebler-Wronski Effect Berlin, Germany (22.4.2009)

2. *Methods for the readout of nuclear spin state silicon*
   Qubit Quantum Information Science and Technology Workshop Berkeley, USA (25.8.2009)

3. *Spin read-out of donors in silicon*
   2009 International Conference on Solid State Devices and Materials Sendai, Japan (7.10.2009)

Joseph Dufouleur

1. *Transport properties of p-doped GaAs nanowires*
   Workshop “Nanowires and Semiconducting Nanotubes, Autrans, France (30.6.2009)

2. *Transport in p-doped GaAs nanowires*
   Seminar ETH Zürich, Switzerland (7.12.2009)

Jonathan Finley

1. *The nuts and bolts of optically active semiconductor quantum dots*
   Tutorial Lectures at “Nanosystems and Sensors” St. Christoph, Arlberg, Austria (3.3.2009)
2. *Structural, physical and optical properties of coupled quantum dot nanostructures*

3. *Optical preparation, manipulation and readout of single electron spins*
   Colloquium Kavli Nanoscience Seminar, TU-Delft, The Netherlands (1.4.2009)

4. *Nanophotonics using semiconductor nanostructures*
   IAS General Assembly, Starnberg, Germany (23.4.2009)

5. *Tunable single quantum dot nanostructures for solid state cavity-QED experiments*
   QIP-2009, Herrsching, Germany (30.6.2009)

6. *Electrically tunable single dot nanocavities*
   International Conference on Modulated Semiconductor Systems (MSS-14), Kobe, Japan (21.7.2009)

7. *Quantum phenomena and dephasing in single dot nanocavities*
   IWPSN International Workshop, Sapporo, Japan (27.7.2009)

8. *Quantum phenomena in cavity QED of single dot nanocavities*
   OECS-11 Conference, Madrid, Spain (8.9.2009)

9. *Quantum dot based nano-photonic system*
   NIM Advisory Panel Workshop, Munich, Germany (31.10.2009)

10. *Optically probing quantum phenomena in artificial atoms and molecules*
    Colloquium Cavendish Laboratory, Cambridge, United Kingdom (4.12.2009)

**José Garrido**

1. *Diamond functional surfaces for bioelectronic applications*
   Seminar, Institut Néel - CNRS, Grenoble, France (12.2.2009)

2. *Diamond transistor arrays for bioelectronic applications*
   Surface and bulk defects in CVD diamond films, XIV, Hasselt, Belgium (2.-4.3.2009)

3. *Diamond science and technology for biosensors and bioelectronics*

4. *Diamond-based platforms for biosensing and bioelectronic applications*
   Seminar, Institute of Toxicology and Genetics, Forschungszentrum Karlsruhe, Germany (28.7.2009)
Alexander Holleitner


2. *Spin relaxation in quasi-one-dimensional electron systems: transition from 2D to 1D* Nanotum Workshop, Technische Universität München, Germany (19.1.2009)

3. *Contacting single nanostructures with electron beam lithography* Raith User Workshop, Dresden, Germany (25.3.2009)

4. *Spin relaxation in quasi-one-dimensional electron systems: the transition from 2D to 1D* Symposium Talk at the DPG Meeting, Dresden, Germany (25.3.2009)


6. *Optoelectronic properties of organic and inorganic nanosystems* Nanosystems, Seminar LS Barth, Technische Universität München, Germany (14.5.2009)


8. *Spatially resolved flow of ballistic electrons by quantized photocurrent spectroscopy* Seminar LS Ensslin/Ihn, ETH Zürich, Switzerland (17.11.2009)

Gregor Koblmüller

1. *Indium nitride (InN): Epitaxial growth, advanced physical properties, and potentials for novel THz emitter* Seminar, Universität Regensburg, Germany (22.1.2009)

2. *Indium nitride (InN): Epitaxial growth, advanced physical properties, and Potentials for novel THz emitter* Seminar, LMU Munich, Germany (20.4.2009)
3. *Indium nitride (InN): Epitaxial growth, advanced physical properties, and potentials for novel THz emitter*
   Seminar, MPI FKF Stuttgart, Germany (25.5.2009)

4. *Indium nitride (InN): Epitaxial growth, advanced physical properties, and potentials for novel THz emitter*
   Seminar, Fraunhofer IAF Freiburg, Germany (26.5.2009)

5. *Indium nitride (InN): Epitaxial growth, advanced physical properties, and effects on defect structure*
   Seminar, UniBW Neubiberg, Germany (22.7.2009)

6. *Influence of structural on electronic properties of high electron mobility core–shell nanowire heterostructures based on InN and InAs*
   DFG-German Research Foundation, Frankfurt, Germany (27.11.2009)

7. *Physics of nanostructured systems Munich: Defect structure and electronic properties in high electron mobility III-V semiconductor materials: from bulk to nanostructures*
   Group Seminar Prof. Kotthaus LMU Munich, Germany (7.12.2009)

**Tillman Kubis**

1. *Predictive quantum theory of current and optical emission in quantum cascade lasers*

**Arne Laucht**

1. *Strong coupling in photonic crystal microcavities*
   Universität Augsburg, Germany (13.5.2009)

**Ralf Meyer**

1. *Frequency-tunable laser diodes for telecommunication applications*
   Institut für Halbleiteroptik und Funktionelle Grenzflächen, Universität Stuttgart, Germany (11.2.2009)
Michael Müller

1. 1.55 µm InP-based short-cavity-VCSEL with enhanced modulation-bandwidth of 15 GHz
   35th European Conference on Optical Communication (ECOC), Vienna, Austria (22.9.2009)

Ulrich Rant

1. Nanorganic, actually!
   Keynote speech at the 2009 General Assembly of the Institute for Advanced Study, Technische Universität München, Prof. Patrick Dewilde, Starnberg, Germany (24.4.2009)

2. Nanorganic, actually!
   Department of Electrical Engineering, Prof. Paolo Lugli, Technische Universität München (28.5.2009)

3. Electrically switchable DNA surfaces for the detection and size analysis of DNA and protein targets
   ECI Conference on Nanoscience and Nanotechnology for Biological/Biomedical/Chemical Sensing, Lanzarote, Spain (14.-17.9.2009)

4. switchDNA biosensors for the label-free detection and sizing of biomolecules on a chip
   Dr. Sam Kavusi, Bosch Research Center, Palo Alto CA, USA (13.11.2009)

5. switchDNA biosensors for the label-free detection and sizing of biomolecules on a chip
   Prof. Nader Pourmand, University of California Santa Cruz, USA (16.11.2009)

6. switchDNA biosensors for the label-free detection and sizing of biomolecules on a chip
   Intel Research Labs, Santa Clara CA, USA (17.11.2009)

7. switchDNA biosensors for the label-free detection and sizing of biomolecules on a chip
   Prof. Z. Siwy, University of California Irvine, USA (20.11.2009)
Reinhard Scholz

1. *Ab initio calculation of the dispersion interaction between a polyaromatic molecule and a noble metal substrate: PTCDA on Ag(110)*
   427th W.E. Heraeus Seminar „Molecular and Organic Electronics: Bridging the Gaps”, Bad Honnef, Germany (25.–29.1.2009)

2. *Spectroscopy on organic semiconductors*
   Oberjoch, Klausurtagung Universität Tübingen, Germany (1.3.2009)

3. *Organic devices: Experimental achievements and challenges for theoretical modelling*
   Universität Würzburg, Graduiertenkolleg 1221, Germany (2.7.2009)

4. *Spectroscopic properties of molecular crystals*
   Universität Würzburg, Graduiertenkolleg 1221, Germany (3.7.2009)

5. *Spektroskopische Eigenschaften von Molekülkristallen und ihren Grenzflächen*
   Technische Universität Ilmenau, Germany (9.7.2009)

6. *Spectroscopic properties of perylene pigments*
   MPI for Polymer Research, Mainz, Germany (29.7.2009)

7. *Spectroscopic properties of molecular crystals: Interference between Frenkel excitons and CT states*
   MPI for Solid State Research, Stuttgart, Germany (6.11.2009)

Ian Sharp

1. *Metal-semiconductor contacts*
   Conference Tutorial, International Conference on Silicon Carbide and Related Materials (ICSCRM-09), Nürnberg, Germany (11.10.2009)

Martin Stutzmann

1. *Biofunktionalisierung von Halbleitern*
   Westfälische Wilhelms-Universität Münster, Allgemeines Physikalisches Kolloquium, Münster, Germany (8.1.2009)

2. *Photovoltaik und Photosynthese*
   TUM Leonardo da Vinci-Zentrum für Bionik, Garching, Germany (22.1.2009)
3. *Nitride versus diamond*
   Ambient GCOE Research Exchange Seminar “Wide Gap Semiconductor; the Presence and its Future”, Waseda University, Japan (2.3.2009)

4. *The Staebler-Wronski-Effect: What we know and what we don’t know…*
   First International Workshop on the Staebler-Wronski-Effect, Helmholtz Zentrum Berlin, Germany (20.4.2009)

Peter Vogl

1. *Giant bias controlled g factor switching in quantum dot molecules*
   International Summer School and Workshop on “Solid State Based Quantum Information Processing - QIP 2009, Herrsching, Germany (3.7.2009)

2. *Novel mesoscopic band structure method for broken-gap superlattices*
   Workshop on Peltier Cooling. Albuquerque, NM, USA (22.4.2009)
8. Courses and Seminars

Lectures

Gerhard Abstreiter

WS 2008/2009  Festkörperspektroskopie  
(zusammen mit Dr. Dominique Bougeard)

SS 2009  Sabbatical

WS 2009/2010  Grundlagen der Halbleiterphysik

Markus-Christian Amann

WS 2008/2009  Werkstoffe der Elektrotechnik  
Technologie der III/V-Halbleiterbauelemente  
(zusammen mit Dr. Ralf Meyer)

Optoelektronik II

SS 2009  Optoelektronik I  
Hauptseminar Optoelektronik  
Praktikum Optische Übertragungstechnik  
(zusammen mit Prof. Erwin Biebl)

WS 2009/2010  Werkstoffe der Elektrotechnik  
Technologie der III/V-Halbleiterbauelemente  
(zusammen mit Dr. Ralf Meyer)

Optoelektronik II
**Dominique Bougeard**

**WS 2008/2009**  
Festkörperspektroskopie  
(zusammen mit Prof. Gerhard Abstreiter)

**SS 2009**  
Halbleiterphysik II – Physik niedrigdimensionaler Systeme

**WS 2009/2010**  
Festkörperspektroskopie

**Martin S. Brandt**

**WS 2008/2009**  
Grundlagen der Halbleiterphysik  
Tutorium zur Halbleiterphysik  
Seminar zu aktuellen Fragen der Magneto- und Spintronik  
(zusammen mit Dr. Sebastian Gönnenwein + PD Markus Betz)  
Seminar "Aktuelle Probleme der Halbleiterphysik"

**SS 2009**  
Halbleiterphysik II – Moderne Bauelemente  
Seminar zu aktuellen Fragen der Magneto- und Spintronik  
(zusammen mit Dr. Sebastian Gönnenwein + PD Markus Betz)  
Seminar "Aktuelle Probleme der Halbleiterphysik"

**WS 2009/2010**  
Einführung in die Festkörperphysik  
Tutorium zur Festkörperphysik  
Seminar zu aktuellen Fragen der Magneto- und Spintronik  
(zusammen mit Dr. Sebastian Gönnenwein + PD Markus Betz)  
Seminar "Aktuelle Probleme der Halbleiterphysik"
Jonathan Finley

WS 2008/2009  Experimental physics I

SS 2009  Experimental physics II

WS 2009/2010  Experimental physics I

José Garrido

WS 2008/2009  Biosensors and bioelectronics I

Seminar “Biosensors and bioelectronics I”

SS 2009  Biosensors and bioelectronics II

NANO 301-1 Electronic Properties of Nanoengineered Materials
Prof. Dr. Gerhard Abstreiter, Prof. Dr. Martin Stutzmann,
Dr. José Antonio Garrido Ariza, together with Dr. Matthew Grayson

WS 2009/2010  Biosensors and bioelectronics

Current topics on functional interfaces
(together with Prof. Martin Stutzmann and Dr. Ulrich Rant)

Alexander Holleitner

WS 2008/2009  Nanosystems I

Nanosystems I – exercise

Seminar: Nanosystems I - physics and instrumentation
(together with Dr. Ulrich Rant)

Seminar über aktuelle Arbeiten der Physik nanostrukturierter Materie
SS 2009  Nanosystems II

Introduction to nanoanalytics
(together with Dr. Helmut Oppolzer)

Seminar: Nanosystems II – physics and instrumentation
(together with Dr. Ulrich Rant)

Seminar über aktuelle Arbeiten der Physik nanostrukturierter Materie

WS 2009/2010  Grundlagen der Experimentalphysik I (LB-Technik)

Nanosystems I
(together with Prof. Jonathan Finley)

Seminar to nanosystems I – physics and instrumentation

Seminar über aktuelle Arbeiten der Physik nanostrukturierter Materie

Gregor Koblmüller

WS 2009/2010  Grundlagen der Halbleiterphysik
(zusammen mit Prof. Gerhard Abstreiter)

Ulrich Rant

WS 2008/2009  Seminar: Nanosystems I - physics and instrumentation
(together with Prof. Alexander Holleitner)

SS 2009  Seminar: Nanosystems II – physics and instrumentation
(together with Prof. Alexander Holleitner)

WS 2009/2010  Current topics on functional interfaces (Seminar)
(together with Prof. Martin Stutzmann and Dr. José Garrido)
Reinhard Scholz

WS 2008/2009  Theoretical physics III: Quantum mechanics

SS 2009  Zentralübung zur Experimentalphysik 2/
  Mathematische Ergänzungen zur Experimentalphysik 2
  Gruppentheorie in Molekülphysik und Festkörperphysik

Martin Stutzmann

WS 2008/2009  Renewable energy I

  Seminar zu "Renewable energy"
  Seminar „Aktuelle Probleme der Halbleiterphysik“

SS 2009  Renewable energy II

  Seminar zu "Renewable energy II"
  Seminar „Aktuelle Probleme der Halbleiterphysik“

  NANO 301-1 Electronic Properties of Nanoengineered Materials
  Prof. Dr. Gerhard Abstreiter, Prof. Dr. Martin Stutzmann,
  Dr. José Antonio Garrido Ariza, together with Dr. Matthew Grayson

WS 2009/2010  Renewable energy I

  Seminar zu "Renewable energy"
  Seminar „Aktuelle Probleme der Halbleiterphysik“

  Biosensors and Bioelectronics
  (together with Dr. José Garrido)

  Current topics on functional interfaces
  (Seminar in Englisch,
   together with Dr. José Garrido and Dr. Ulrich Rant)
Peter Vogl

WS 2008/2009  Theoretical semiconductor physics I

SS 2009  Theoretische Festkörperphysik

WS 2009/2010  Theoretical Semiconductor Physics I
Walter Schottky Seminars

*Formation and doping of quantum dots studied at the atomic scale by cross-sectional scanning tunneling microscopy*
Prof. Paul Koenraad, COBRA-Department of Applied Physics, Eindhoven University of Technology, The Netherlands (13.1.2009)

*Indistinguishable photons from the resonance fluorescence of a single quantum dot in a microcavity*
Serkan Ates, Institut für Halbleiteroptik und Funktionelle Grenzflächen, Universität Stuttgart, Germany (16.1.2009)

*Coherent optical control of semiconductor quantum dots*
Dr. Andrew Ramsay, University of Sheffield, UK (20.1.2009)

*Ultrafast, nonlinear, and quantum nanolasmonics*
Prof. Mark I. Stockman, Department of Physics and Astronomy, Georgia State University, Atlanta, Georgia, USA (23.1.2009)

*Photonic crystals: Properties, modeling and applications*
Prof. Wolfgang Freude, Institute of High-Frequency and Quantum Electronics (IHQ) Karlsruhe, Germany (27.1.2009)

*Selective excitation of single spins in a nanowire quantum dot*
Maarten van Weert, Kavli Institute of Nanoscience Delft, Delft University of Technology, The Netherlands (3.2.2009)

*Phonon dephasing in semiconductor quantum dots: Entangled photon sources and cavity QED*
Prof. Ulrich Hohenester, Institut für Physik, Fachbereich Theoretische Physik, Karl-Franzens-Universität Graz, Austria (9.2.2009)

*Spin dynamics in semiconductors: Manipulating spin by electric fields*
Prof. Peter Schwab, FB Physik, Universität Augsburg, Germany (10.2.2009)

*Classical and quantum information processing using photonic crystals*
Dr. Dirk Englund, Harvard University, USA (10.3.2009)

*Radiative cascades in semiconductor quantum dots*
Prof. David Gershoni, Technion – Israel Institute of Technology, Haifa, Israel (24.4.2009)

*Optical generation of microwave signals*
Prof. Ning Hua Zhu, Institute of Semiconductors, Chinese Academy of Sciences Beijing, China (12.5.2009)
Designing and characterizing semiconductor lasers using strain
Prof. Alf Adams, Photonics Group, Advanced Technology Institute, University of Surrey, UK (15.6.2009)

Non-blinking semiconductor nanocrystals: Suppression of nonradiative Auger processes
Dr. Alexander Efros, Naval Research Laboratory, Washington DC, USA (16.6.2009)

Solitonic microlasers and polarization competition in broad-area vertical-cavity surface-emitting lasers
Dr. Thorsten Ackemann, SUPA and Department of Physics, University of Strathclyde, Glasgow, Scotland, UK (19.6.2009)

Quanteneffektbauelemente für Wärmebildkameras – Farbe im Infrarot
Dr. Martin Walther, Fraunhofer-Institut für Angewandte Festkörperphysik IAF Freiburg, Germany (23.6.2009)

Deterministically positioned quantum dot subject to electric fields in any crystal orientation
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