Preface

Report writing usually belongs to the more unpleasant activities in the life of a productive scientist. Indeed, in many cases there is lot of reason to doubt that the considerable effort invested in the compilation of scientific reports has any noticeable effect at all. Fortunately, judging from the generally positive feedback we have received so far each year, this does not seem to be the case for our Annual Report.

Therefore, all members of the Walter Schottky Institut are pleased to present to you this Annual Report 2002 and sincerely hope that you will find it interesting and informative. As customary, most of the space is devoted to extended abstracts describing ongoing research projects. Almost all of these brief project summaries have been written by the PhD or Diploma students directly involved in the work, and it is their enthusiasm and dedication which we hope to convey to you. In addition, relevant statistical and bibliographical information has been compiled at the end for convenient reference.

The year 2002 has been one of considerable changes and reorientation. As a major new research direction, biochemical and bioelectronic applications of semiconductors by now have been firmly established at the Walter Schottky Institut. This has been made possible by a considerable amount of funding from the Deutsche Forschungsgemeinschaft (Sonderforschungsbereich 563), the Bavarian Research Foundation (FORNANO), and from industry (Fujitsu Europe). New activities have also been started in the field of spintronics and semiconductor-based quantum computation. Of course, more conventional topics of basic semiconductor physics and device technology still continue to play the most important role in our research activities and currently account for about 80% of our annual budget. The successful launch of VERTILAS GmbH as a spin-off company of the Institute demonstrates that innovative basic research on one hand and commercial product development on the other hand are compatible even for small research institutions.

2002 has also been a year of major changes as far as senior members of the scientific staff of the Institute are concerned. Early in the year, Arthur Zrenner and Oliver Ambacher have left us to assume a position as Full Professors at the Universities of Paderborn and Ilmenau, respectively, Christoph Nebel has accepted a new appointment as a permanent senior staff member in the Physics Department of the TU München, and Karl Brunner just has left to the University of Würzburg, where he has taken a tenured position as Associate Professor. We wish all of them good luck and success and hope that they will keep pleasant memories of their time at the Walter Schottky Institut.

Let me finally take this opportunity to express my sincere gratitude to everyone who has contributed to and supported our work directly or indirectly in the last year. You all have helped to create the environment and atmosphere necessary for successful scientific work. Thank you!

Garching, April 2003

Martin Stutzmann
1. The Walter Schottky Institut

History

The Walter Schottky Institut (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 the 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, InSb, GaN/AlGaN, SiGe, and Si/SiO₂ based heterostructures (MBE, CBE, plasma-induced MBE)
- 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
- Hall measurements and magneto-transport
- FTIR spectroscopy
- X-ray photoelectron spectroscopy
- High frequency parameter analyzers
- Electron spin resonance (ODMR, EDMR)
- He³ cryostats with magnetic fields up to 15 and 17 Tesla
- Special characterization facility for laser diodes

Computational facilities:
- High end workstations
Research groups

The WSI accommodates four research groups headed by Gerhard Abstreiter, Markus-Christian Amann, Martin Stutzmann and Peter Vogl, with a total headcount of about 90 including scientific and technical staff, secretaries, and doctorate as well as diploma (master) students. Out of these, about 25 permanent positions are authorized by TUM, while basically all the doctorate positions are financed via research projects with external funding. The main research interests are:

- fabrication and characterization of new semiconductor materials, material combinations, as well as hetero- and nanostructures
- basic physics with emphasis on electronic and optical properties of low dimensional systems
- realization of new semiconductor devices for application in ultrafast electronics, optoelectronics and sensors
- theory and simulation of modern semiconductor materials and devices

Experimental Semiconductor Physics I (Gerhard Abstreiter):

Research projects of this group 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 of analytical tools for the characterization of nanometer-size structures, 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 the development of novel infrared optical detectors and charge storage based on quantum dots, new concepts for unipolar intersubband lasers, the control and manipulation of charge and spin in quantum devices like single dot photodiodes, and the test of semiconductor nanostructures as chemical/biological sensors.

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 antimonide 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 sub-
strates, InGaAsP and AlInGaAs vertical cavity surface-emitting laser diodes at 1550 nm and microcavity light-emitting devices in the 1300 nm wavelength range. Recent work also covers the development of wavelength-tunable laser diodes for wavelengths above 2 μm using antimonide 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):**

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 nano structuring, 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 constitute
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 devices and of low dimensional semiconductor structures.
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Zero field spin splitting in nitrides and their heterostructures

Jacek A. Majewski and Peter Vogl

Recently, it has been predicted that diluted magnetic semiconductor GaMnN can be ferromagnetic even at room temperature. This makes GaMnN a very promising candidate for a source of spin polarized carriers and opens new perspective for spin devices entirely based on nitrides. The possible design of such spintronic devices requires the understanding of the spin dynamics and spin relaxation processes, which are, in turn, determined by the zero field spin splitting of the bands. Unfortunately, the spin splitting of the bands, induced by the spin orbit interaction, is unknown in nitride heterostructures. Particularly important issue is the role of huge electric fields in the nitride heterostructures, since it is known that the electric field causes spin splitting (so-called Rashba effect). The spin splitting of the bands can be considered as the result of an effective lateral $k$-dependent magnetic field that induces a coherent spin precession of the spin polarized carriers. In this report, we present the studies of the spin splitting in bulk AlN, GaN, and AlN/GaN superlattices. These studies are based on first-principles, relativistic local density calculations that provide very accurate description of the spin splitting in nitride bulks. For example, the calculated spin splitting of the valence bands in $\Gamma$ point in cubic ($\Delta_0$) and wurtzite ($\Delta_1$ and $\Delta_2$) GaN agrees excellently with magneto-optical experimental data, i.e., $\Delta_0=18.6$ (18) meV, $\Delta_1=5.3$ (5.4) meV and $\Delta_2=6.7$ (5.8) meV with experimental values given in parenthesis. Here we concentrate on the lowest order in $k$ spin splitting, i.e., terms linear in $k$. For wurtzite bulks and superlattices grown along the $c$-axis, the effective 2x2 Hamiltonian describing $k$-linear terms can be written in the form $H = \sigma \cdot B_{\text{eff}}$ with $B_{\text{eff}}(k_x, k_y) = a_R(k_x, -k_y, 0)$, where $\sigma$ is the vector of Pauli matrices, $k_x$ and $k_y$ are wave vector components in the plane perpendicular to the hexagonal axis, and $a_R$ is a band dependent constant that determines the isotropic spin splitting energy $\Delta E_{\text{spin}} = 2 a_R (k_x^2 + k_y^2)^{1/2}$ of a certain band. In the wurtzite structure, the combination of crystal-field and spin-orbit interactions leads to a three-edge structure of the top of the valence band in the $\Gamma$ point, known as A, B, and C edges. Two of these three edges are of $\Gamma_7$ and one of $\Gamma_9$ symmetry, while the lowest conduction state has $\Gamma_7$ symmetry. The valence states are, in order of decreasing energy, $\Gamma_9$, $\Gamma_7$, $\Gamma_7$ for GaN, and $\Gamma_7$, $\Gamma_9$, $\Gamma_7$ for AlN (i.e., like in ZnO). It is known that the $\Gamma_9$ states of wurtzite are not linearly split. In Fig. 1 we depict calculated values of the constant $a_R$ for $\Gamma_7$ valence and conduction band states in AlN and GaN. For comparison, Fig. 1 contains the values of $a_R$ constant for few other compounds crystallizing in wurtzite structure. As can be seen in Fig. 1, in all wurtzite semiconductors, the linear-$k$ spin splitting is considerably smaller in the conduction band than in the valence band. In further discussion, we focus on the spin splitting of the conduction band in AlN/GaN superlattices. We have performed calculations for AlN/(GaN)$_n$ superlattices with various width of the GaN quantum

Fig. 1: Linear-$k$ spin splitting constant [in eVÅ] for valence and conduction band of wurtzite structure semicon...
well, measured by the number of GaN layers \( n \). The divergence of the large electric polarization (spontaneous and piezoelectric) causes a negative and positive bound sheet charges located at the interfaces, which leads to huge electric fields in the structure of the order of few MV/cm. A schematic band lineup of \((\text{AlN})_4/(\text{GaN})_8\) superlattice together with the calculated wave functions of the top of the valence band and bottom of the conduction band are shown in Fig. 2. The wave function of the valence band top is strongly localized at one interface, just giving rise to the negative polarization charge.

The calculated values of the constant \( \alpha_R \) for \( \text{AlN}/(\text{GaN})_n \) superlattices are depicted as a function of the quantum well width in Fig. 3. The magnitude of \( \alpha_R \) in superlattices increases in comparison with the value in the bulk GaN \( (\alpha_R = 0.007 \text{ eVÅ}) \) and saturates for quantum wells consisting of about 16 GaN layers, reaching the value \( \alpha_R = 0.0125 \text{ eVÅ} \). The \( k \)-linear terms dominate spin splitting in the Brillouin Zone around \( \Gamma \)-point. For \( k_v = 0.25 \text{Å}^{-1} \) (roughly \( \frac{1}{4} \) of the BZ), the linear term in spin splitting energy is 6meV out of 8meV of the total spin splitting. It is clear that the spin splitting can be observed only in samples with large electron density (large Fermi wave vector). Recently, a Japanese group has found \( \alpha_R = 0.018 \text{ eVÅ} \) from magnetoresistance oscillations of the 2DEG in AlGaN/GaN heterostructures, which agrees fairly well with our calculations. Altogether, it shows that the conduction band spin splitting of the AlN/GaN heterostructures remains small in spite of the huge electric fields, indicating long spin relaxation times in the nitride heterostructures. For comparison, the \( \alpha_R \) for [111] AlAs/GaAs heterostructure is by factor 10 larger (equal to 0.14 eVÅ). On the other hand, the spin splitting of the conduction band can be only marginally influenced by the external electric field. The calculations show that the external bias of 250 kV/cm changes the constant \( \alpha_R \) by approximately 1%. These results suggest that GaN heterostructures are a promising candidate for some spintronic applications, where long spin lifetime is needed. GaN quantum wells are not particularly suitable for spin transistor (rather weak steering possibility) but they can be very good for static Qubits. In this aspect, the GaN heterostructures grown along non polar directions, which have recently achieved considerable attention, can be even better, since the spin splitting in such structures is by a factor 10 smaller than in the heterostructures grown along polar [0001] direction.

**Fig.2:** Band lineup and wave functions of the valence (VBT) and conduction band (CBB) edges. For both bands, the laterally and macroscopically (envelope) averaged first-principles wave-functions are shown.

**Fig.3:** Linear-\( k \) spin splitting constant \( \alpha_R \) for AlN/(GaN)_n superlattices.
Spin precession in superlattices and quantum wells of III-V materials

Jacek A. Majewski and Peter Vogl

The present surge in interest for spin devices and spin relaxation effects calls for a detailed understanding of the spin dynamics in semiconductor nanostructures. We will focus on [001] grown zincblende structures in this report, where the spin-split energies in the conduction bands may be considered as the result of an effective lateral $k$-dependent magnetic field. This field induces a coherent spin precession of incoming electrons which is counteracted by incoherent spin-independent and spin-dependent scattering events. These spin-related effects have been studied extensively both theoretically and experimentally, yet there are still many open questions concerning the microscopic origin of the spin-splitting of the energy bands, their dependence on material combinations, on quantum well widths of heterostructures, and on electric fields. Therefore, we have attempted to answer some of these questions by first-principles, relativistic local density calculations, augmented by empirical tight-binding studies. Here, we focus on the spin splitting energies of the lowest conduction band. We find that band-structure engineering allows one to tailor the spin precession frequency of conduction electrons in nanostructures by very significant amounts. Zincblende heterostructures or superlattices with a [001] growth axis possess a symmetry that leads to a Mexican-hat-type structure of the conduction band edge near the $\Gamma$-point. This is a consequence of a $k$-linear spin splitting of the conduction band states of the form

$$\Delta E_{\uparrow\downarrow} = 2k_z \sqrt{\alpha_{BIA}^2 + \alpha_{SIA}^2} - 2\alpha_{BIA} \alpha_{SIA} \sin(2\theta),$$

where the components of the lateral wave vector perpendicular to the growth axis are given by $k_x = k_y \cos \theta, k_y = k_z \sin \theta$. For structures that possess tetragonal $D_{2d}$ symmetry, only the so-called bulk inversion asymmetry (BIA) coupling $\alpha_{BIA}$ is nonzero. In systems of lower $C_{2v}$ symmetry that support a polar axis (such as asymmetric quantum wells or no-common-atom superlattices), the so-called structure inversion asymmetry (SIA) or Rashba $\alpha_{SIA}$ term leads to an angular modulation in the spin-splitting energy. In bulk zincblende materials, there are no $k$-linear terms in the conduction band at the $\Gamma$-point. A symmetric heterostructure A:B:A possesses a nonzero SIA coupling if the materials A and B share no common atom since the interface asymmetry destroys the $D_{2d}$ symmetry in such a case. We have carried out LDA calculations of $(\text{InP})_n:(\text{AlSb})_m$ and $(\text{InP})_n:(\text{GaAs})_m$ short-period superlattices and compared them to common-atom $(\text{AlAs})_n:(\text{GaAs})_m$ superlattices for various values of $n$ and $m$. We find, for example, $\alpha_{SIA} = 0.112$ eVÅ and $\alpha_{SIA} = 0.043$ eVÅ for the first conduction band of the $(\text{InP})_3:(\text{AlSb})_3$ and $(\text{InP})_3:(\text{GaAs})_3$ superlattices, respectively. Detailed studies lead us to conclude that the SIA coupling is primarily determined by the wave-function asymmetry within the quantum well layers but is insensitive to the spin-orbit interaction at the interface or within the barrier layers. The BIA constants are given by $\alpha_{BIA} = 0.034$ eVÅ, $\alpha_{BIA} = 0.064$ eVÅ, $\alpha_{BIA} = 0.052$ eVÅ, for $(\text{AlAs})_3:(\text{GaAs})_3$, $(\text{AlSb})_3:(\text{InP})_3$, and $(\text{InP})_3:(\text{GaAs})_3$, respectively, thus having the same order of magnitude for all of these superlattices. In Fig. 1 we show a comparison of the square of the laterally averaged electronic wave function of the lowest conduction band at the $\Gamma$-point of a common-atom $(\text{AlAs})_3:(\text{GaAs})_3$ and no-common atom $(\text{AlSb})_3:(\text{InP})_3$ superlattice, respectively.
The asymmetry in the wave function throughout the quantum well region in the latter case is clearly seen. Note that the sequence of atoms across the AlAs/GaAs structure is Al-As-Ga-...-Ga-As-Al and therefore mirror-symmetric, in contrast to the AlSb/InP superlattice with Sb-Al-Sb-(In-P)n-Al-Sb that is not mirror-symmetric for any integer n.

The laterally averaged square of the zone center conduction band wave function in (a) common atom (AlAs)3(GaAs)3, and (b) no-common-atom (AlSb)3(InP)3 superlattices.

Any reduction of the D_{2d} symmetry by an applied electric field or by an asymmetric strain in common-atom heterostructures leads to a finite SIA coupling in principle. We show here that an applied electric field affects both BIA and SIA terms in an unexpected way. We study common-atom AlAs:(GaAs)n: AlAs heterostructures as a function of n and assume that an applied potential drops linearly within the GaAs well region but remains zero in the (doped) barrier regions. In Fig. 2 we show the calculated BIA- and SIA-coupling associated with the lowest conduction band edge for up to n = 100 GaAs layers and for three different electric fields F. For narrow heterostructures, \( \alpha_{BIA} \) is seen to be independent of F in accord with simple envelope function theory. For the wider wells, however, the potential drop reaches almost 1 eV which is smaller than but of the order of the band width of the lowest conduction band. This leads to a localization of the conduction band wave function in real space that is shown in Fig. 2 for the various field values and for the case n = 50. Correspondingly, it becomes delocalized and a superposition of many band states in \( k \)-space. This is a situation that clearly goes beyond simple effective mass theory. As a consequence, we find this distorted wave function to lead to a BIA coupling that increases approximately linearly with the electric field and saturates for long well widths. This saturation can easily be understood: for a given field value F, the wave function is localized at one side of the quantum well region and does not feel a prolongation of the well in the opposite direction. The SIA coupling shows exactly the same trend as the BIA coupling.

![Spin splitting constant vs Width of QW](image1)

![Square of the laterally averaged conduction band wave function at the \( \Gamma \)-point](image2)

Fig. 1: The laterally averaged square of the zone center conduction band wave function in (a) common atom (AlAs)3(GaAs)3, and (b) no-common-atom (AlSb)3(InP)3 superlattices.

Fig 2. (a) Calculated BIA and SIA constants of the conduction band in AlAs/GaAs heterostructures. (b) Square of the laterally averaged conduction band wave function at the \( \Gamma \)-point.

supported by DFG (SFB 348 (A4))
Spin wave resonance in Ga$_{1-x}$Mn$_x$As

S. T. B. Goennenwein$^1$, T. Wassner, M. S. Brandt, M. Stutzmann, A. Koeder, W. Schoch, and A. Waag

Magnetic resonance is a particularly powerful method to investigate the properties of ferromagnetic materials. When spin waves or magnons are observed with this technique, the effective ferromagnetic spin stiffness $D=2A/M$ can be directly deduced, where $A$ is the exchange constant, and $M$ the saturation magnetization. Obviously, such measurements are of particular interest when the magnetic properties of novel materials are being investigated. Such a material is the dilute magnetic semiconductor Ga$_{1-x}$Mn$_x$As, which makes it possible to take advantage of ferromagnetic properties in devices made exclusively from semiconductors.

The Ga$_{1-x}$Mn$_x$As samples studied here were grown at 250 °C by molecular beam epitaxy on semi-insulating (001) GaAs substrates at the Universität Ulm. The ferromagnetic resonance was measured in a conventional electron spin resonance system at 9.3 GHz using magnetic field modulation. Figure 1 shows the ferromagnetic resonance spectra of different samples obtained at $T=5$ K, with the external field $B_0$ perpendicular to the sample surface. For $x=0.022$ and $x=0.025$, rather broad resonances with indications of substructure are obtained. For $x \geq 0.036$, up to ten distinct resonances are resolved. The position of the resonances $B_{\text{res},n}$ is indicated schematically by the arrows for the topmost spectrum. For convenience, the different modes have been enumerated with the index $n$, starting with $n=0$ for the highest $B_{\text{res}}$ observed.

![Fig.1: Ferromagnetic resonance spectra of Ga$_{1-x}$Mn$_x$As samples with different Mn content $x$ at $T=5$K. The external magnetic field was oriented perpendicularly to the sample surface. The spin wave resonances are indicated schematically by arrows in the uppermost spectrum and enumerated with the index $n$.](image)

The lowest spectrum shows the paramagnetic resonance of Mn$^{2+}$ ions in GaAs with a Mn concentration [Mn]$<0.01\%$, as can be clearly seen by the characteristic hyperfine lines.

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arising from the $^{55}\text{Mn}$ nucleus with $I=5/2$. We identify the multiple FMR resonances for $x\geq 0.036$ with spin wave resonances (SWR). However, the resonances observed do not follow the classical pattern for spin waves. Conventionally, the quadratic dispersion relation for magnons leads to a variation of the mode spacing according to $(B_{\text{res},0} - B_{\text{res},n}) \propto n^2$. As shown in more detail in Fig. 2, this is incompatible with the experimentally observed dependence. It is well established that e.g. a gradient in magnetization or in uniaxial magnetic anisotropy can lead to such a non-quadratic mode spacing.

Independent of its specific origin, such a gradient in the magnetic properties can be translated into an effective magnetic field $B_{\text{eff}}$. For $B_{\text{eff}} = \Delta B z$ varying linearly with depth $z$ across the film thickness, one has $(B_{\text{res},0} - B_{\text{res},n}) = D^{\frac{2}{3}}(\Delta B 2\pi/3)^{2/3}(n+1/4)^{2/3}$, in very good agreement with the experimentally observed dependence. Fitting the data with this relation thus allows to determine the spin stiffness $D=2A/M$, provided $\Delta B$ is known. In turn, $\Delta B$ can be obtained e.g. from measurements of films with different thicknesses. Accordingly, we have performed a series of wet etching experiments for the sample with $x=0.051$, thereby successively reducing the thickness of the Ga$_{1-x}$Mn$_x$As film. The FMR spectra are found to indeed strongly depend on film thickness. Considering $B_{\text{res},0}$ as an estimate for $B_{\text{eff}}$, one indeed finds $B_{\text{eff}} = \Delta B z$, with $\Delta B \approx 3.5$ mT/nm. With the saturation magnetization $M = (2.3 \pm 0.3) \times 10^4$ A/m as determined by superconducting quantum interference device (SQUID) dc magnetization measurements, one finally finds $A = (1.5 \pm 0.6)$ pJ/m for $x=0.051$. This value agrees well with theoretical calculations, which have yielded $0.2 \text{pJ/m} < A < 1.2 \text{pJ/m}$ for $x=0.044$.

The microscopic origin of the linear gradient in the magnetic properties observed here is currently subject to several investigations. Most probably, a variation of the hole density $p$, possibly due to a varying density of compensating defects like As antisites or interstitials, is at the origin of the magnetic gradient. It is therefore imperative to accurately control both the Mn content $x$ and the hole density $p$ if one is to tailor the magnetic properties of Ga$_{1-x}$Mn$_x$As for specific applications.

Fig.2: Comparison of the experimentally observed spin wave mode spacing (full circles) with theoretical predictions (straight lines). Clearly, the quadratic spacing expected for classical spin waves is incompatible with the experiment, while the spacing corresponding to a linear gradient in magnetic properties agrees well with the measurements.
Properties of the high-spin ground state of Mn$^{2+}$ in GaN

Tobias Graf$^\dagger$, Mario Gjukic, Martin Hermann, Oliver Ambacher, Martin S. Brandt, and Martin Stutzmann

Because of their potential use in spintronic devices, transition metal doped dilute magnetic semiconductors (DMS) currently attract significant scientific interest. For GaN:Mn as one of the most promising materials, room temperature ferromagnetism was predicted for Mn$^{2+}$ and hole concentrations around $10^{21}$ cm$^{-3}$. In last year’s report, we have summarized optical investigations of GaN:Mn, which lead to the conclusion that Mn$^{3+}$ is a deep state not suitable for long-range ferromagnetic coupling. Here, we look in more detail into the microscopic properties of Mn$^{2+}$ spins in MBE-grown GaN:Mn films by electron spin resonance (ESR) spectroscopy.

According to Hund’s rule, the spin and orbital angular momentum of the five Mn$^{2+}$ d-electrons are expected to couple to a $6A_1$ high-spin ground state with the electronic spin $S = 5/2$ and the angular momentum $L = 0$. In the presence of an external magnetic field $B$, the degeneracy of this ground state is lifted by the Zeeman interaction $\mathcal{H}_{\text{Zeeman}} = \mu_B B g \hat{S}$ with $g = 2.000$, as shown in Fig. 1. An axial fine structure splitting of the five otherwise equal ESR transitions with $\Delta m_s = 1$ is described in the spin Hamiltonian by $\mathcal{H}_{\text{Fine}} = D \left( S_z^2 - \frac{1}{3} \hat{S} (\hat{S} + 1) \right)$ with an axial crystal field parameter $D = -2.7$ µeV. Such a deviation of the Mn$^{2+}$ spin states from spherical symmetry can be related to the distortion of the neighboring atom configuration from a perfectly tetrahedral lattice in hexagonal GaN.

The configuration of these neighboring atoms can be obtained from the hyperfine interaction $\mathcal{H}_{\text{Hyperfine}} = \hat{S} \hat{A} \hat{I}$, which enters the spin Hamiltonian via the Fermi contact interaction of the Mn$^{2+}$ electronic spin with the nuclear spin $I = 5/2$ of the natural isotope $^{55}$Mn. This interaction would be zero for pure d-states. However, core polarization, admixture of excited s-like states, and covalent bonding contribute, to a net Fermi contact interaction of $A = -0.8$ µeV in GaN. This value is very similar to literature values of $A$ in other materials with anion electronegativities close to that of nitrogen, e.g. for substitutional Mn$^{2+}$ in ZnS or ZnO. Therefore, it can be concluded, that Mn$^{2+}$ occupies a nearly substitutional Ga site in GaN. Because of the competition between $\mathcal{H}_{\text{Zeeman}}$, $\mathcal{H}_{\text{Fine}}$, and $\mathcal{H}_{\text{Hyperfine}}$, the Mn$^{2+}$ spin states can be approximated by pure $|m_s, m_i\rangle$ Eigenstates of $S_z$ and $I_z$ only for an orientation $\Theta = 0^\circ$ of the c-axis of the GaN crystal along the external magnetic field. Otherwise, significant level mixing is observed experimentally, as well as in the exact numerical solution of the full spin Hamiltonian, as shown in Fig. 2. For orientations with $10^\circ > \Theta > 80^\circ$, the

\[ \mathcal{H}_{\text{Spin}} = \mathcal{H}_{\text{Zeeman}} + \mathcal{H}_{\text{Fine}} + \mathcal{H}_{\text{Hyperfine}} \]

Fig. 1:
Spin splitting and resonant transitions of the Mn$^{2+}$ ground state with $S = 5/2$ and $I = 5/2$.
contribution of "forbidden" transitions with $\Delta m_i > 0$ in the central fine structure group is very strong and cannot be accounted for by perturbation theory any more.

However, for the Mn concentrations of our samples around $10^{20}$ cm$^{-3}$, no other para-

magnetic or ferromagnetic resonance features besides the signature of isolated, paramagnetic Mn$^{2+}$ centers could be detected by ESR.

Because $L = 0$ for a pure $^6A_1$ ground state, theoretical calculations of $D$ must include spin-orbit coupling up to high order perturbation theory. However, the calculated values are still orders of magnitude smaller than those observed experimentally. To explain this discrepancy, a small displacement of the Mn$^{2+}$ ion from the tetrahedral Ga lattice site has been postulated, but has not been confirmed independently yet because of the lack of other microscopic experimental data.

A quantitative verification of this displacement is possible from ESR data of Mn$^{2+}$ in strained GaN films with the superposition model of crystal fields. As shown in Fig. 3, the observed dependence of the axial crystal field parameter $D$ of the GaN layer strain is in good agreement with the weak-field superposition model for a displacement of 8.5 pm along c from a substitutional Ga site.

**Fig. 2:**
Measured (dots) and calculated (lines) ESR transition fields for GaN:Mn$^{2+}$. The six "allowed" hyperfine lines for the five fine structure groups are mixed at intermediate orientations, as marked at $\Theta \sim 10^\circ$ for the central fine structure group.

**Fig. 3:**
Dependence of the axial crystal field splitting of Mn$^{2+}$ on the strain in different GaN layers. The experimental observations agree well with the predictions from the superposition model of crystal fields.

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Theory of vertical and lateral Stark shifts of excitons in quantum dots

M. Sabathil*, S. Hackenbuchner, S. Birner, J. A. Majewski, P. Vogl, and J. J. Finley

The shape and alloy composition of self-assembled InGaAs quantum dots that are buried beneath a cap layer of GaAs is currently of great interest but still poorly understood. There are several direct experimental methods such as STM, TEM, or x-ray diffraction that provide some information about the composition of the quantum dots. However, all of these techniques have the drawback of either being destructive or being limited to dots on the surface that differ from buried quantum dots (QD). By contrast, the recent development of single dot spectroscopy allows one to examine the optical properties of individual quantum dots under the influence of external perturbations such as electric or magnetic fields. The dependence of the optical and electronic properties on these external fields provides an important albeit indirect probe of the shape and alloy composition of the quantum dot. This situation calls for the development of a reliable, quantitative theoretical model for the electronic properties of QD of any given shape and alloy profile, in order to be able to link measured exciton energies or relative transition intensities of single QD with their shape and composition.

In this work we present systematic predictions of the bias dependence of the electronic structure of self-assembled quantum dots for a wide variety of dot shapes, alloy profiles and show that the combined effect of applied and (piezoelectric) internal electric fields yields detailed information about shape and composition profiles. We therefore chose a multiband-$\mathbf{k} \cdot \mathbf{p}$ Schrödinger-Poisson approach that takes into account the Coulomb interaction between confined electrons and holes, as well as the applied and induced piezoelectric fields fully self-consistently. The present calculations have been carried out with the device simulator nextnano$^3$. Based on single particle eigenstates obtained from the solution of an 8-band $\mathbf{k} \cdot \mathbf{p}$ Hamiltonian, the exciton energy is calculated self-consistently within the Hartree approximation. Assuming a separable exciton wave function, the Coulomb interaction between electron and hole is calculated by iteratively solving the Poisson and Schrödinger equation for each particle, taking into account external and internal potentials including image charges due to the variation of the dielectric constant. The equations are solved within a finite differences scheme on an inhomogeneous grid with 1 nm grid spacing inside the quantum dot resulting in about $3 \times 10^5$ grid points. The strain is calculated by minimizing the total elastic energy, including the wetting layer and a sufficiently large GaAs substrate and cap volumes, in order to minimize artefacts from boundaries. The resulting strain-

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induced band deformations and piezoelectric charges are fully taken into account in the above self-consistent cycle. In this way we have calculated exciton transition energies for various In(Ga)As dot shapes and alloy profiles under the influence of vertically and laterally applied electric fields.

The simulated structure consists of a 20 nm thick GaAs substrate, a 1 nm wetting layer with an Indium content of 50 % and the quantum dot with an In concentration that varies from 50 % to 100 %. The dot is capped by a GaAs layer. We have examined two classes of QD shapes that have been discussed in the literature, lens-shaped and truncated pyramidal ("obelisk") dots namely [Fig. 1]. The height of the QD has been systematically varied from 3 to 5 nm and the base width from 15 to 25 nm, respectively. These values lie within the range of experimentally estimated dot sizes. In addition, we considered two types of alloy profiles, a simple linear and an angular profile. In the latter case, the Indium concentration varies with the polar angle relative to the center axis of the dot [Fig. 2]. This profile is supported by STM measurements and by theoretical models of the growth process and leads to an inverted pyramid of high Indium content inside the quantum dot. The most important feature of this alloy profile is the fact that it produces a vertical and a lateral variation of the Indium concentration which has been neglected so far.

The polarizability of the exciton in a lateral electric field is sensitive measure of the alloy profile of the quantum dot that depends particularly on the base width of the QD [Fig. 3]. In the case of the vertically linear alloy profile, one has a laterally homogeneous Indium concentration. Not surprisingly, this yields a very large polarizability for both types of QD shapes. The radial alloy profile, on the other hand, leads to an inverted pyramid of Indium content in the QD that in turn gives a strong lateral confinement of the hole. This effect reduces the polarizability drastically, typically by a factor of two compared to the linear alloy profile. Therefore, the alloy profile can be determined once the width of the quantum dot is known. The dots with a radial profile show a base width dependence of the polarizability that is very similar for both types of QD shapes. By contrast, QD with a laterally homogeneous profile possess a strong width dependence of the polarizability when they are obelisk-shaped but a weak one in the lens-shaped case. This is due to the fact that the curved surface of the lens QD enhances the lateral confinement which in turn limits the polarizability.

![Fig. 2. The angular alloy profile where the Indium concentration depends on the polar angle relative to the center axis of the QD. Light and dark grey-scale indicates high and low In content, respectively.](image2)

![Fig. 3. Predicted lateral polarizability shown as a function of the base width of the quantum dot for different shapes and alloy profiles. The polarizability is much larger for dots (A) with a laterally homogenous alloy profile than for (B) dots with a radial profile since the latter provides a strong lateral confinement for the hole. For both alloy profiles, the results are shown for obelisk-shaped (full lines) as well as for lens-shaped quantum dots (dashed lines).](image3)
Rabi-oscillations of an exciton in a single self-assembled quantum dot

Evelin Beham¹, Artur Zrenner, Frank Findeis, Max Bichler, and Gerhard Abstreiter

We report the demonstration of coherent optical writing combined with an electrical read-out of a single quantum state, which could be used as a quantum bit (QUBIT) in the context of future quantum information technologies. Our electrical approach favorably allows for a quantitative detection of the prepared quantum state and hence for the transfer of the QUBIT-information in an electrical signal.

Whilst based on a conventional GaAs n-i-Schottky diode, the only optically active part is a single self-assembled InGaAs quantum dot (QD) contained in the intrinsic layer of the diode (see inset of Fig. 1). The optical isolation of a single QD is achieved using near-field shadow masks. A single QD embedded in such a diode structure can be treated as a quantum mechanical two-level system. Experimentally, we apply a photocurrent (PC) absorption technique to investigate the coherent behavior of the QD and we were able to directly observe Rabi-oscillations within the excitonic QD two-level system.

The exciton resonance is characterized by linear, incoherent PC spectroscopy. The basic concept of our PC technique is shown by means of the band diagram in the inset of Fig. 1. A QD is embedded in the intrinsic layer of the photodiode that is operated in reverse bias. Resonant optical excitation of the QD ground state is performed by a cw Ti:sapphire laser tuned to the exciton resonance. In a regime of sufficiently high electric field \( F \), the optically generated carriers tunnel out of the QD and are separated in the applied field. The resulting field ionization of the carriers gives rise to a tunneling current, which is recorded in an external dc current measurement. By tuning the excitation energy, the PC signal exhibits sharp spectral resonances revealing the discrete density of states of a single QD (Fig. 1).

A single QD can be regarded as a quantum mechanical two-level system. The upper level is given by the QD occupied by one exciton in its ground state. The lower level represents the QD without any occupation, which is the crystal ground state (for a schematic view see inset of Fig. 2). For the discussion of the coherent properties of our photodiode we compare the relevant timescales in this system. The lifetime of a coherent polarization is given by the dephasing time \( \tau_p \), the characteristic time to observe coherent interactions in the system. Recent experiments on self-assembled InGaAs QDs have evidenced low temperature dephasing times in excess of 500 ps. The spontaneous lifetime of the exciton \( \tau_s \) is of the order of 1 ns and naturally the longest timescale for the excitonic two-level system. Furthermore, with respect to our PC technique, the tunneling time \( \tau_t \) is of strong relevance. Depending on the electric field, \( \tau_t \) is tunable from infinity down to <1 ps.

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Tunneling on a timescale $\tau_t < \tau_p$ on the other hand leads to enhanced dephasing and limits the time range for coherent interactions. In order to obtain conditions for coherent excitation and avoid unwanted coherent dynamics we apply 1.7 ps pulses from a mode-locked Ti:sapphire laser. By applying an appropriate bias on our photodiode, we choose a sufficiently long tunneling time that exceeds the pulse length but is smaller than the dephasing time.

If a quantum mechanical two-level system is strongly driven at its resonance frequency, there occurs a basic coherent phenomenon, known as Rabi-oscillations. For time scales shorter than the dephasing time, the population of this two-level system oscillates between the lower and the upper state. Basic theory describes this oscillation of the excitation level $L$ by $L = \sin^2(\Theta/2)$, where the parameter $\Theta = \mu \int E(t) dt$ contains the transition dipole moment $\mu$ and the envelope area $A = \int |E(t)| dt$ of the excitation pulse. This envelope area $A$ determines the resulting inversion after pulsed excitation, determined by both the pulse length and its electric field amplitude $A_{\text{exc}}$. In our experiments we use a fixed pulse length and vary the pulse intensity $I \sim A_{\text{exc}}^2$.

To demonstrate the coherent properties of our single QD photodiode we performed PC spectroscopy under pulsed excitation. The pulsed laser is spectrally set to 1.31 eV, in resonance to the exciton transition. Figure 2 shows the resulting PC as a function of excitation amplitude $A_{\text{exc}}$. With increasing $A_{\text{exc}}$ we first observe an increasing PC signal from the QD. At $A_{\text{exc}} \approx 1$ the current reaches a maximum, a further increase of $A_{\text{exc}}$ in turn leads to a reduction of the PC. This behavior is in contrast to the well known characteristics of a conventional, incoherent photodiode, but expected for the coherent population of a two-level system, which is directly reflected here in the PC. With increasing $A_{\text{exc}}$ we observe damped Rabi-oscillations in the PC reflecting quantitatively the occupancy of the two-level system. The first maximum corresponds to a $\pi$-pulse excitation appearing here at $A_{\text{exc}} \approx 1$. Applied to a two-level system with an initial occupancy of 0, a single $\pi$-pulse leads to a complete inversion of the system (occupancy 1). The photo-ionization of this excitation leads to the separation of (in the ideal case) exactly one electron-hole pair and to the net transport of one elementary charge between the contacts of the photodiode. Hence, a single QD photodiode excited by $\pi$-pulses is a deterministic current source delivering one elementary charge $e$ to an outer circuit per laser pulse. With the pulse repetition frequency $f = 82 \text{ MHz}$ of our laser we expect a time-integrated net current $I = fe$ of 13.1 pA. The Rabi-oscillation shows a maximum PC of about 11.5 pA nearly reaching the theoretical PC maximum. This quantitative correspondence makes our experiment novel in the field of Rabi-oscillations. In contrast to most other approaches, the applied PC technique provides a quantitative proof for Rabi-flopping.

![Fig. 2: Rabi-oscillations of the PC at resonance for increasing excitation amplitude $A_{\text{exc}}$. Coherent $\pi$-pulse excitation corresponds to $A_{\text{exc}} = 1$. The symbols show the experimental data, the line is included to guide the eye.](image)
Optical interband Hofstadter Butterfly

C. Strahberger and P. Vogl

In 1967, R. D. Hofstadter discovered that the motion of an electron that is subject to a two-dimensional periodic potential as well as a magnetic field with a cyclotron radius that is comparable with the lattice constant shows interesting and unusual commensurability effects. Whenever the magnetic flux through the unit cell is a rational number, the effect of the magnetic field is to act like a periodic supercell potential and leads to a band structure with energy gaps. This band structure depends on the periodicity of the potential which changes quasi-continuously with the magnetic flux. Consequently, one finds a fractal eigenspectrum of the electrons as a function of the magnetic field.

So far, it has been very difficult to observe this Hofstadter butterfly because the energy spectrum of the 2D electron system typically lies in the far infrared. Here, we show theoretically that this effect can be observed in the visible spectrum via interband magneto-optics.

Consider an undoped GaAs quantum well that is covered by a thin AlGaAs cap layer. On top of this structure, a periodic array of stressors or metallic gates can produce a periodic potential that influences the valence and conduction band edge states in the quantum well (Fig. 1). One can tune these stressors or gates to produce a weak periodic potential leading to minibands of at least 50 meV width in the conduction band and 10 meV in the valence band (Fig. 2).

Fig. 1: The potential in the 10 nm GaAs channel is periodically modulated by a two-dimensional array of stressors or gates. The effective lattice constant of this periodic potential modulation has been assumed to be 10 nm.

Fig. 2: Schematic energy band diagram of the periodically modulated 2D-channel structure depicted in Fig. 1. Shown are the highest valence and lowest conduction subband, respectively, as a function of the lateral wave vector.

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In Fig. 3, we show the predicted absorption coefficient as a function of the magnetic field, where we have used full valence and empty conduction states at zero temperature. The photon energy on the vertical axis is given in eV. The absorption strength is indicated by a logarithmic gray scale color code; the darker the dots, the higher the absorption coefficient. One can clearly see the fractal spectrum that is characteristic of the Hofstadter butterfly.

According to envelope function theory, the allowed transitions are interband intra-Landau transitions with $\Delta n = 0$ with cyclotron energies that contain the reduced electron and hole mass. Since the hole mass is 5 times higher than the electron mass, the whole absorption spectrum reflects predominantly the conduction band spectrum. Indeed, the $0 \rightarrow 0$ and $1 \rightarrow 1$ transitions clearly dominate the spectrum for small fields. Already for relative magnetic flux values of 0.1, however, we find forbidden transitions with $\Delta n = 1$ to gain oscillator strength. For higher fields, the absorption spectrum mimics the fractal Hofstadter eigenvalue spectrum, again with some noticeable exceptions: the absorption coefficient is negligible around the middle of the band for large magnetic flux values, whereas the density of states shows a dense maximum in this region. This is a consequence of the s- and p-type character of the states which leads to molecular selection rules that reflect the symmetry of the lattice.

In summary, we have shown that the optical absorption spectrum corresponding to the Hofstadter butterfly generally mimics its fractal energy spectrum but reveals many additional features that are not present in the density of states. For low magnetic fields, one recovers the standard selection rules for magneto-absorption. whereas for high fields, these selection rules are strongly broken. We predict that the Hofstadter butterfly can be directly observed optically via interband absorption in a periodically gated heterostructure.

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Probing the structure of integer quantum Hall edges with momentum-resolved tunnel spectroscopy

Michael Huber¹, Matthew Grayson, Dieter Schuh, Max Bichler, and Gerhard Abstreiter

In a quantum Hall system the edge is of particular interest as the confining edge potential together with the perpendicular magnetic field causes the electrons to drift along the sample boundary creating a chiral current. The one-dimensionality of these edge channels gives rise to quantized conductance irrespective of the shape of the edge potential. Therefore in transport measurements the inner structure of the quantum Hall edge, which strongly depends on the sharpness of the confining edge potential, remains hidden. At a moderately soft edge potential this structure is supposed to depend on correlation effects of interacting electrons. A very smooth edge is characterized by the formation of wide compressible stripes that can be described in a single particle picture, while at an infinitely hard potential wall the confining edge potential is supposed to dominate over any interaction energy. In this work we studied the structure of quantum Hall edge states at an edge potential that is compatible with the predicted appearance of correlation effects and we identify features that cannot be explained with a simple non-interacting edge picture.

Using cleaved-edge overgrowth we created a device with two orthogonal quantum wells (QW-A and QW-B) separated by a T-shaped tunnel junction (inset of Fig.2). A high magnetic field perpendicular to QW-A creates quantum Hall edge states close to the T-junction which are probed with QW-B acting as a tunnel contact. Due to the high interface quality at the cleaved edge the geometry of the sample causes strict momentum conservation for tunneling charges along the barrier. Resonances in the tunnel conductance correspond to coincidences of electronic states in the energy-momentum space. With the magnetic field transverse to the tunnel current the Lorentz force acting on the tunneling electrons shifts the dispersion in k-space and the voltage bias shifts the energy, allowing

Fig.1: (Sample 1) Color plot of differential tunnel conductance $dI/dV$ as a function of bias voltage $V$ and magnetic field $B$ (red represents high and blue low conductance). Negative bias indicates tunneling of electrons into the quantum Hall edge. Solid lines indicate maxima in the tunnel conductance. In these resonance maxima we can identify the dispersion of the edge states of individual Landau levels as well as the dispersion curve of the 2D tunnel contact (dashed line).

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us to map out an entire $E$ vs. $k$ dispersion relation. We are able to resolve the dispersion of both the integer quantum Hall edge modes and the 2D Fermi surface (Fig. 1) With the edge state dispersion reflecting the shape of the edge potential, we have a direct measure of the electrostatics in a quantum Hall edge. Electrostatic simulations of the device under negative bias quantitatively confirm the identification of conductance peaks with spin-unresolved Landau levels (Fig. 1). We have already proven the capability of this method as reported last year. Here we focus on the low bias regime. Extrapolating this understanding to the zero bias conductance trace shows which peak corresponds to which filling factor. Momentum conservation results in a tunnel conductance $G \sim \int_{E} t_{k} A_{QHE}(k,E) A_{2D}(k,E) \, dk$ proportional to a convolution of the spectral functions of both the 2D Fermi contact $A_{2D}(k,E)$ and the quantum Hall edge $A_{QHE}(k,E)$ with a broadening $t_{k}$ of the resonances. While the spectral function of the quantum Hall edge is a $\delta$-function the 2D contact yields van-Hove singularities. Analyzing $t_{k}$ within this model we are able to exclude lifetime broadening of the resonances due to scattering in the plane of the 2D contact, which would leave its mark in Lorentzian shaped resonances. Our experiment however features gaussian broadening, which we attribute to inhomogeneities and scattering at the tunnel barrier. Fig. 2 shows the good agreement of our model with the experiment concerning positions and shape of the main resonance peaks. There is however an additional resonance in the shoulder of the lowest Landau level that is seen in all of the samples studied so far and which cannot be explained within our non-interacting edge state model. At zero bias we are probing electronic states at the Fermi energy, where correlation effects should play a role in the energetic structure at the edge either through the exchange enhanced spin-splitting (present also in the bulk), or through edge reconstruction induced by Coulomb interactions. Either of these two effects is therefore a reasonable candidate for explaining this feature. An exchange enhanced spin gap would increase the separation of the edge modes of the two spin orientations. The peak offset would then be a measure of the exchange energy. In the edge-reconstruction model, the peak offset would define the Fermi momentum of the outermost Fermi-point of the reconstructed edge. Measurements with an additional in-plane magnetic field could help to clarify this open question by separately addressing spin related effects.

**Fig. 2:** (Sample 2) Differential tunnel conductance $dI/dV$ at zero bias. Resonances due to tunneling into spin degenerate Landau levels are indexed, and explain the main structure except for a secondary peak in the shoulder around 4.1 T. Inset shows a schematic section through the sample: a T-shaped combination of two high-mobility quantum wells separated by a tunnel barrier.

work performed in collaboration with W. Biberacher, Walther-Meissner-Institut, Garching (Germany); supported by DFG (Schwerpunktprogramm Quanten-Hall-Systeme)
The corner-junction quantum well: A new device geometry for studying the quantum Hall effect

Matthew Grayson, Michael Huber, Frank Fischer, Dieter Schuh, Max Bichler, and Gerhard Abstreiter

We demonstrate a new type of quantum confined structure consisting of a high-mobility GaAs/AlGaAs heterostructure overgrown on top of a precleaved corner. This corner-junction quantum-well heterostructure (CQW) exhibits high mobility with fractional quantum Hall effect (QHE) signatures on both facets, allowing new geometries for integer and fractional QHE edge experiments to be realized.

To regrow simultaneously over multiple crystal facets, the facets must be of the same crystal class, otherwise the quality of the growth will be compromised. To prepare orthogonal \{110\} class facets, one can either take a (001) wafer and make two orthogonal cleaves on the (110) and (1-10) facets, respectively (Fig 1a.), or one can take a (110) ‘substrate’ wafer and cleave it along its (1-10) ‘precleave’ facet (Fig 1b.) The two types of pieces, which we call ‘square’ (Q) and ‘long’ (L), respectively, can then be mounted on a tantalum cleaved-edge overgrowth block and be simultaneously regrown as shown in the lower part of Fig. 1. The sample is overgrown with a standard heterojunction structure, but with all growth fluxes (Ga, Al, As), all growth thicknesses (Ga, Al) and all shutter times (Si) increased by a factor of \(\sqrt{2}\) to compensate for the 45° tilting of the substrate. The resulting sample is contacted with indium on the sample edges away from the overgrown corner to make electrical contact.

In Fig 2, we confirm the sharpness of the resulting morphology in a test superlattice structure. The dark bands consist of 100 nm thick layers of AlAs separated by lighter GaAs/AlGaAs bands.

Fig. 3 shows the 4-point longitudinal resistance, \(R_{xx}\), vs. the normal component of the magnetic field for two different overgrown facets of an L-type CQW sample (05-23-02.4L3). The facets show different densities and different mobili-

Fig.1:
Q- and L-type corner quantum well pieces as mounted on the tantalum block for growth. \(\theta\) defines the angle of tilted magnetic field during measurement.

Fig.2:
Scanning electron micrograph of corner overgrowth.
ties, as evident from the position and strength of the $R_{xx}$ minima, respectively. The (110) substrate facet in the bottom trace has a lower density ($n = 1.07 \times 10^{11}$ cm$^{-2}$) than the (1-10) precleave facet in the upper trace ($n = 1.30 \times 10^{11}$ cm$^{-2}$) although the substrate shows a higher mobility with its stronger 2/3 fractional minimum.

This new structure is particularly advantageous for abrupt junctions of coupled quantum Hall systems with different filling factor. According to the equation inset in Fig. 4, the ratio of filling factors in the (110) substrate, $\nu_s$, and the (1-10) precleave, $\nu_p$, can be tuned by choosing the appropriate angle for the tilted magnetic field. In particular, in tilted fields oriented between $\theta = +45^\circ$ and $+135^\circ$ as defined in Fig. 1b), the edge states at the junction are co-propagating. Such a situation is impossible to realize in standard planar quantum wells. We demonstrate the expected equilibration through such co-propagating edge states in Figure 4, where a 4-point measurement of the Hall resistance across the junction is shown to be equal to the sum of the Hall resistances within each system.

**Fig.3:**
$R_{xx}$ for the (110) and (1-10) facets on the L-type regrown CQW (sample 05-23-02.4L3).

**Fig.4:**
Quantized resistance plateaus:

$R_{xy} = h/ne^2$ with $n$ listed in the figure.

Dashed curves: $R_{xy}$ for the (110) and (1-10) facets on one of the L-type regrown CQW's.

Solid curve: $R_{4pt}$ showing new resistance quanta equal to the sum of the constituent $R_{xy}$'s.
Millikelvin temperature measurements of high-mobility (110)-GaAs heterostructures

Frank Fischer¹, Matthew Grayson, Erwin Schuberth*, Dieter Schuh, Max Bichler, and Gerhard Abstreiter

Recently there has been increasing interest in new types of proposed ground states within high Landau levels of the quantum Hall effect (QHE). The charge density waves (CDW) or striped phases are experimentally observable through transport anisotropies of the longitudinal resistance at half-odd-integer fillings from 9/2 upwards. But up to now it is unclear why these states always align along a specific crystal direction. The two main prerequisites to observe these effects are high mobility 2-dimensional electron systems (2DES) with \( \mu > 9 \times 10^6 \text{ cm}^2/\text{Vs} \) and temperatures lower than 100 mK. The systems normally under investigation are (001) oriented GaAs/AlGaAs modulation-doped heterostructures (e.g. WSI Annual Report 2001). There, the stripes are aligned along either the [110] or [1-10] direction depending on electron density or in-plane magnetic fields.

To find out more about the effect which is breaking the symmetry, we use a different substrate orientation, namely (110) GaAs/AlGaAs \( \delta \)-doped heterostructures with a spacer thickness of 800 Å resulting in peak-mobilities up to \( \mu = 4 \times 10^6 \text{ cm}^2/\text{Vs} \) at densities of around \( n = 2.1 \times 10^{11} \text{ cm}^{-2} \). Since it is not known if surface anisotropies due to the growth are somehow pinning the striped phases, it is necessary to grow in a way that the (110) surface is as smooth and rotationally invariant as possible. But, due to the non-polar nature of the surface, this is no trivial problem. We were able to achieve growth parameters resulting in almost completely smooth surfaces with a low defect density, as shown in Fig. 1 with sample 06-20-2.3. In comparison sample 08-20-02.1 with different growth parameters shows triangular defects pointing in the [001] direction typical for (110) growth.

<table>
<thead>
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<th>Sample</th>
<th>06-20-02.1</th>
<th>08-20-02.1</th>
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</tbody>
</table>

Fig. 1: Comparison of surface morphology and growth parameters of 2 growths made at different substrate temperatures and As4 beam equivalent pressures (BEP). Scale is 420 \( \mu \text{m} \times 320 \mu \text{m} \).
Fig. 2 shows a pre-characterization of sample 06-20-02.1 at \(T = 320\text{mK}\). The fractional states 2/3 and 3/5 are very well developed, proving that the sample indeed is of very high quality.

Due to the fragile nature of the anisotropic ground states it is necessary to measure at very low temperatures, especially in the case of (110) 2DEG’s where the mobility is not as high as in (001) based systems. These measurements are being done in collaboration with the Walther-Meissner-Institut for low temperature physics in a dilution refrigerator cryostat with a base temperature of about 5mK and a magnetic field up to 6T.

A trace of the longitudinal resistance for two perpendicular current directions at \(T = 8\text{mK}\) is shown in Fig.3. The fractions from 9/2 upwards show a highly anisotropic behavior, similar to the one already seen in (001) systems. The temperature-dependence seen in Fig. 4 shows an increase of longitudinal resistance for the [1-10] direction and a decrease for the [001] direction. Basically no anisotropy is observed for the 7/2 resistance peak. However there is a striking anisotropy developing for lower fractions, especially below filling factor 2, which may lead to new physics not related to the behaviour for high filling factors.

Fig. 4 shows the temperature-dependence of the 9/2-maximum for the two perpendicular current directions shown in Fig. 2.

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Three-dimensional electron gases in AlGaN/GaN heterostructures

A. Link¹, O. Ambacher, and M. Stutzmann

The experimental realisation of three-dimensional electron slabs (3DES) was achieved by tailoring polarisation charges in the III-V nitride semiconductor system [1]: Instead of a AlGaN/GaN heterojunction which induces a two-dimensional electron gas, a graded AlGaN/GaN transition over a distance of 100nm by a linear change of the Al-content was used. This spreads the positive 2D polarisation charge into a bulk 3D polarisation background charge (see sample structure in Fig. 1). The charge profile is given by the divergence of the polarisation field, which changes only along the (growth) z-direction: 

\[ N_{P}^{Pol}(z) = \nabla \cdot P = \frac{\partial P(z)}{\partial z}. \]

The linear grading of the Al-content and thus the polarisation results therefore in an approximately uniform charge profile given by 

\[ N_{P}^{Pol}(z) = \frac{(P(z_0) - P(0))}{z_0}. \]

P(z₀) is the spontaneous and piezoelectric polarisation of AlₓGa₁₋ₓN for the local Al-composition at z = z₀. This fixed background charge attracts carriers from remote donor-like states. The mobile three-dimensional electron slab thus formed behaves similar to bulk doped carrier systems. But compared to bulk-doped heterostructures the electrons in the polarisation-doped 3DES exhibit a higher mobility due to the absence of shallow donors and do not freeze out at low temperatures. These characteristics allow us to observe quantum effects in low temperature magnetotransport measurements. In particular, the 3DES shows clear Shubnikov-de Haas (SdH) oscillations. A study of these oscillations reveals the effective mass of carriers, the collisional broadening due to scattering, and the transport mechanisms in the polarisation-doped 3DES. The angular dependence of the SdH-oscillations displays the three-dimensional character of the carrier gas.

The measured Hall mobility of the 3D electrons is determined to \( \mu_H \approx 3000 \text{cm}^2/\text{Vs} \) at T = 400mK, which is much higher than in conventional bulk doped systems. Assuming that the 3DES is spread over a thickness d, the sheet carrier density of the 3DES is calculated to \( n_{3DES} \cdot d = 7.2 \cdot 10^{12} \text{ cm}^{-2} \). The thickness d of the 3DES is obtained from a self-consistent Schrödinger-Poisson band calculation as d = 75nm, due to 25nm depletion of the 3DES by the surface potential. This depletion in the graded AlGaN layer has been verified by capacitance-voltage profiling. Thus the Hall 3D carrier density is

\[ \text{Fig. 1: Schematic band diagram and charge distribution in the 3DES heterostructure} \]

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\( n_{3\text{DES}} \approx 10^{18} \text{cm}^{-3} \). This value is confirmed by the density deduced from SdH measurements. As shown in Fig. 2, the measurement of \( R_{xx} \) shows pronounced oscillations in a magnetic field range above 3T. The frequency of these oscillations in 1/B is related to the carrier density \( \Delta (1/B) = \frac{e\hbar}{m^*\varepsilon_F} \sim (n_{3\text{DES}})^{2/3} \), which thus can be determined to \( n_{3\text{DES}}(\text{SdH}) = 1.1 \cdot 10^{18} \text{cm}^{-3} \). The value of the averaged effective mass \( m^* = 0.19m_0 \) used therein was extracted from the temperature dependent decrease of the SdH minima. For comparison, the theoretically predicted value of the effective mass of electrons in an Al\(_{0.11}\)Ga\(_{0.89}\)N alloy is 0.21\(m_0\), so the measured electron mass in the graded alloy seems reasonable.

From the exponential decay of the SdH amplitude with decreasing magnetic field a quantum scattering time of \( \tau_q = 0.3\text{ps} \) was extracted. The classical scattering time calculated from the Hall mobility \( \mu_H = e\tau_c/m^* \) is \( \tau_c = 0.34\text{ps} \). Thus within the experimental error the ratio \( \tau_c/\tau_q \) is close to unity. Since the classical scattering time is weighted by a scattering angle factor \( (1-\cos\Theta) \), while the quantum lifetime is determined by all scattering events, the ratio of both can be used to identify the dominant scattering mechanism in the 3DES. At low temperatures the most important scattering mechanisms that can affect mobility are alloy disorder scattering, charged dislocation scattering, and ionised impurity scattering from the remote donors at the surface states. Since both last-mentioned mechanisms strongly favour small angle scattering and therefore would cause a \( \tau_c/\tau_q \) ratio much larger than unity, they can be excluded from playing a dominant role. If alloy disorder scattering is assumed to be the most important mobility limiting factor, the short range scattering potential \( V_0 \) can be estimated from the scattering rate: \( \tau_{\text{alloy}}^{-1} \sim V_0^2x(1-x)g_{3\text{D}}(\varepsilon_F) \), where \( x \) is the alloy composition and \( g_{3\text{D}}(\varepsilon_F) \) the three-dimensional density of states. Matthiessen’s rule was used for a spatial averaging of the scattering rate to describe the graded alloy. To obtain a mobility of \( \mu = 3000\text{cm}^2/\text{Vs} \), an alloy scattering potential of \( V_0 = 1.8\text{eV} \) is necessary. For comparison, the conduction band offset between the binaries forming the alloy is 2.1eV.

Furthermore, angle resolved SdH investigations were performed to prove the 3D character of the electron slab. Fig. 2 shows SdH measurements with the magnetic field parallel to the growth direction and gradually tilted towards the sample surface. The oscillation frequency remains almost invariant with tilt angle \( \alpha \). The slight shift of the SdH minima may arise from an anisotropy of the effective electron mass.


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Point defects at the interface between high-\(k\) oxides and silicon

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Thermal Si\(\text{O}_2\) shows excellent properties as a gate dielectric. Nevertheless, the continuous scaling down of the gate insulator layer thickness in advanced metal-oxide-semiconductor (MOS) devices is expected to reach fundamental limits in the near future. One possible solution is to use gate insulators with a dielectric constant higher than that of Si\(\text{O}_2\), such as Al\(\text{2O}_3\) and Zr\(\text{O}_2\). This should enable the use of physically thicker oxide layers, which could lead to a reduction in leakage current and to improved gate reliability. For this reason, such alternative gate insulators are currently being investigated for future generations of MOS transistors. In addition to their high dielectric constants, Al\(\text{2O}_3\) and Zr\(\text{O}_2\) are attractive because of their thermal stability in contact with Si and large Si/oxide barrier heights. An important issue of these oxides is the microstructure of the point defects present at their interface to Si, which can be crucial for the device performance as shown by previous studies on the Si/Si\(\text{O}_2\) system. We have used electrically detected magnetic resonance (EDMR) to study the defects at the interface between these high-\(k\) dielectrics and Si(100) substrates.

Four different types of interfaces have been investigated, obtained by atomic layer deposition (ALD): Si(HF last)/Al\(\text{2O}_3\), Si/Si\(\text{O}_2\)/Al\(\text{2O}_3\), Si(HF last)/Zr\(\text{O}_2\) and Si/Si\(\text{O}_2\)/Zr\(\text{O}_2\), where HF last indicates that the native oxide has been removed prior to the oxide growth. The change of photoconductivity induced by spin resonance was detected with a lock-in amplifier using a magnetic-field modulation under steady-state illumination with blue light. The spectra obtained on Si(HF last)/Al\(\text{2O}_3\) for different orientations of the sample with respect to the magnetic field \(H\) are shown in Fig.1.

The spectra in Fig.1 appear to be the superposition of an isotropic signal and an anisotropic one. The best deconvolutions have been obtained using a center with trigonal symmetry.

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together with an isotropic center and Lorentzian lines with \( g_\perp = 2.0074 \pm 0.0003 \) and \( g_\parallel = 2.0024 \pm 0.0003 \) for the trigonal center, and line widths \( \Delta H_{pp\perp} = 9.3 \pm 0.6 \text{G} \) and \( \Delta H_{pp\parallel} = 6.2 \pm 0.5 \text{G} \). A similar dependence of the line width has been reported previously and partially attributed to strain present at the interface. For the isotropic signal, we found \( g = 2.0055 \pm 0.0003 \) and \( \Delta H_{pp} = 7.9 \pm 0.3 \text{G} \), common to dangling bond centers observed in disordered and amorphous silicon. The trigonal center is responsible for about 60% of the total EDMR signal intensity.

In Fig. 2, we show the angular dependence of the g-factors obtained from the deconvolution together with the g-factor anisotropy of the the Pb0 center reported by Stesmans et al. on the Si(HF last)/Al2O3 interface, obtained with conventional EPR measurements. In our case, the difference between \( g_\perp \) and \( g_\parallel \) is smaller than the one reported by Stesmans. We can also fit our data with the superposition of the known Pb0 center and an isotropic center. However, the intensity of the isotropic signal increases significantly in this case. In Fig. 2, we also show the g-factors obtained from the deconvolution of the spectra obtained on the Si/SiO2/Al2O3, Si(HF last)/ZrO2 and Si/SiO2/ZrO2 interfaces. The Si/SiO2/Al2O3 interface shows essentially the same g-factors as the Si(HF last)/Al2O3 interface. The angular dependence of the g-factor of the trigonal center obtained for the Si(HF last)/ZrO2 and Si/SiO2/ZrO2 interfaces is closer to the one reported by Stesmans.

The observed similarity between the Si/SiO2 and the Si(HF last)/Al2O3 interface is intriguing considering that the latter interface has been claimed to be atomically abrupt. However, the similarity becomes less surprising taking into account recent results based on Z-contrast scanning transmission electron microscopy revealing a thin (< 0.3 nm) SiOx interface layer even in HF treated samples and the strongly localized nature of the Pb center wave-function. In contrast to the Si(HF last)/Al2O3 interface, the similarity between the Si/SiO2 and the Si(HF last)/ZrO2 interface is expected because of the known difficulties in obtaining abrupt interfaces.

**Fig. 2:** Angular dependence of the g-factors observed for defects at the Si/oxide interface. The dots are the experimental results, while the solid lines show the anisotropy of the trigonal center and the isotropic line (dangling bond). The dashed lines show the anisotropy of the Pb0 center reported by Stesmans et al. Included in the figure are the results of the deconvolution of the Si/SiO2/Al2O3, Si(HF last)/ZrO2 and Si/SiO2/ZrO2 interfaces.
Light-induced modification of amorphous silicon suboxides (a-SiOₓ)

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The structural, electronic, optical and electrical properties of amorphous silicon suboxides (a-SiOₓ) have been investigated with respect to light-induced degradation and laser crystallisation. The samples were deposited by plasma enhanced chemical vapour deposition (PECVD) from the source gases SiH₄, H₂ and CO₂ and contain oxygen fractions of 0 - 44 at. % at a hydrogen concentration of roughly 20 at. %. Their large-scale and cheap production makes them interesting candidates for optoelectronic and photovoltaic applications.

The hydrogen, usually present in amorphous silicon-based layers, helps to passivate Si dangling bond defects which have negative effects on the optical and electrical properties of the material. However, when subjected to strong illumination, additional defects are created, which is known as the so called Staebler-Wronski effect in amorphous silicon (a-Si:H). As a consequence, the dark and the photoconductivity are significantly reduced. These changes are metastable and reversible. The exact underlying mechanisms, though, are still unclear and widely discussed. The a-SiOₓ:H samples ([O] = 0 - 44 at. %) used in this study of metastability were subjected to above-bandgap illumination at an intensity of 350 mW/cm² for 16 hours and annealed for one hour at 170 – 200 °C to investigate the reversibility. No effect of this degradation procedure on the optical gap E₀₄ could be noticed, whereas a metastable increase of Urbach energy and disorder at higher oxygen contents (> 30 at. %) could be detected. The defect density of all samples with [O] = 0 - 44 at. % revealed a similar increase of metastable dangling bonds upon light-soaking. Correspondingly, the photoconductivity σₚʰ decreased as a function of the illumination time. As an example, fig. 1 displays the time evolution of the defect density and the photoconductivity of an a-SiOₓ:H sample with 30 at. % oxygen during illumination. The number of dangling bonds, as determined from the midgap absorption, increases according to a power law Nₐ ∝ t⁻⁰·₂. Simultaneously, a comparable decay of σₚʰ ∝ t⁻⁰·₂ is observed. The behaviour of both quantities is qualitatively similar to the findings for a-Si:H (Nₐ ∝ t⁻¹/₃ and σₚʰ ∝ t⁻¹/₃), although light-induced degradation seems to proceed more slowly in a-SiOₓ:H. All the light-induced effects described above can be reversed by

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annealing at 200 °C for one hour. Thus, reversibility of the Staebler-Wronski-effect also exists in the case of amorphous hydrogenated silicon suboxides.

The manufacturing of solar cells and also flat panel displays requires large areas of high quality semiconductor layers, available at low costs. Microcrystalline silicon produced by laser crystallisation is one feasible approach to this technological problem. In our study H-effused silicon suboxides with 0 – 44 at. % [O] were crystallised homogeneously and by means of two-beam interference by 8 ns tripled Nd:YAG laser pulses (λ = 355 nm). The melting threshold I_{thr} of the different samples was determined by the transient change in reflection which occurs when a strong laser pulse is able to melt the sample surface. I_{thr} decreases linearly from above 80 mJ/cm² for a-Si:H to approximately 40 mJ/cm² at oxygen contents above 40 at. %. The intense laser pulses produced an additional shoulder at about 1.1 – 1.2 eV (indirect band gap of c-Si) in the absorption spectra. This lead to the assumption that a phase separation in Si crystallites and a surrounding amorphous suboxide matrix occurs upon laser irradiation, which was further corroborated by transmission and reflectivity measurements. Fig. 2 displays atomic force micrographs of an a-SiOₓ sample with 40 at. % oxygen which was irradiated by two interfering laser beams. At low pulse energy density (fig. 2 a), this results in a smooth line pattern with 1 µm period and a roughness below 10 nm. However, at increased laser energy (fig. 2 b), spots with a height of roughly 25 nm appear along the crystallised lines and provide further support for a phase separation.

Raman measurements of the crystallised SiOₓ samples confirmed the existence of Si crystallites. Amorphous silicon suboxides, especially at oxygen contents of [O] ≤ 40 at. %, tend to form silicon crystallites similar to pure amorphous silicon, although their composition is more complex. However, above an oxygen threshold of 40 at. %, no Si crystallites could be formed by means of strong laser pulses. This is probably a consequence of the increasing oxygen content in the samples which prevents an efficient formation of μc-Si inclusions. The observed phase separation in Si crystallites and SiOₓ matrix might turn out to be interesting for a (holographic) structuring of the samples and for applications in the permanent high density storage of data.

Fig. 2: Atomic force micrographs and cross section analysis of an a-SiOₓ sample with 40 at. % oxygen, crystallised by means of two-beam interference at laser energy densities of 60 (a) and 120 mJ/cm² (b).
Aluminum-induced crystallization of silicon germanium alloys

Mario Gjukic¹, Michael Buschbeck, Robert Lechner and Martin Stutzmann

As worldwide energy resources decrease, alternative concepts for the supply of energy become more important. In addition to water and wind power, photovoltaics represent a promising energy source for the future. For example, state-of-the-art crystalline silicon solar cells reach efficiencies up to 25 percent. However, to compete with today’s energy sources like nuclear and fossil power, the costs of photovoltaic devices have to be reduced drastically.

There are two main reasons for the high production cost of efficient crystalline silicon solar cells. Firstly, they require an absorption layer with a thickness in the range of several ten µm, since the direct band gap does not fit well to the solar irradiation spectrum. Secondly, high quality monocrystalline substrates are expensive which also leads to an increase of cost.

To lower the price of crystalline solar cells one has to use cheap substrates as well as semiconductors with improved absorption behavior. Therefore, the deposition of silicon or silicon germanium (SiGe) on glass is of increasing scientific and commercial interest. However, glass as a substrate is not suitable for epitaxial growth of crystalline materials. After the deposition of SiGe on glass, the semiconductor is amorphous. Suitable methods like Solid Phase Crystallization (SPC), Laser Induced Crystallization (LIC) or Metal Induced Crystallization (MIC) have to be applied to achieve micro- or polycrystalline materials with good optical and electronic properties.

In this project, we investigate the promising technique of Aluminum-Induced Crystallization of Si and Ge. Fig. 1a shows the initial sample structure after the deposition of Al and a-Si on glass. The annealing process, which causes a layer exchange and the crystallization of the Si layer is carried out below the eutectic temperature of the Si/Al system ($T_{E, Si/Al} = 566^\circ C$). Shortly after the temperature treatment starts, Si nuclei begin to form at the Al/a-Si interface (Fig. 1b). The crystallites grow into the Al layer until they reach the glass substrate (Fig. 1c). At this point lateral growth starts (Fig. 1d) and leads to the formation of a polycrystalline silicon layer (Fig. 1e).

Fig.1: Schematic cross-section view of an aluminum-induced layer exchange (ALILE) process: (a) – initial sample structure; (b) to (e) – temporal evolution of the annealing process

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Fig. 2 shows optical micrographs obtained for a glass/Al/Ge structure. The left picture demonstrates the temperature dependence of the process with the annealing time kept constant at 4 h; the right picture shows the final state after 20 h at the highest temperature ($T = 390 \, ^\circ\text{C} < T_{E,\text{Ge/Al}} = 420 \, ^\circ\text{C}$). The size of the formed crystallites reaches values up to 110 µm in the case of Ge and 70 µm in the case of Si for a film thickness of only several hundred nm.

A small part of the Al remains in the polycrystalline Si/Ge layer and acts as a shallow acceptor. Thus, the semiconductor shows p-type conductivity. We determined the carrier concentration by Hall measurements. For a fully crystallized poly-Si layer the typical hole concentration is $3 \cdot 10^{18} \, \text{cm}^{-3}$ and the hole mobility 40 cm$^2$/Vs.

Depending on the charge carrier concentration the layer can either be used in photovoltaic devices as active absorber material, as back surface field layer or as a metallic back-contact. Also the use as a seed layer for epitaxy is another possible application.

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Raman spectroscopy of self-assembled Si/Ge quantum dots

Dominique Bougeard, PingHeng Tan, Matthias Sabathil, Hubert Riedl, Karl Brunner, and Gerhard Abstreiter

Structural parameters and the band structure properties of localised hole states in small Ge quantum dots (QDs) on Si are studied in detail by phonon and electronic Raman spectroscopy on QD ensembles. The investigated samples were grown by self-assembly during solid-source molecular beam epitaxy on n-(100) Si substrates at 510 °C. The QD layers contain 8 monolayers of Ge. In AFM measurements of uncapped samples the QDs were found to have a lateral size of about 20 nm and a height of 2 nm. The areal density is about 1.5x10¹¹ cm⁻². The studied samples contain 80 QD layers which are separated by 25 nm Si. All the structures are capped with 100 nm Si.

Structural parameters of the studied nanostructures as the Ge content and strain were studied by phonon Raman scattering because they affect the spectral position and the relative intensity of the SiGe lattice vibrational modes. A typical spectrum of the phonon modes of an as-grown Si/Ge QD structure is shown in the topmost spectrum of Fig. 1. After subtraction of the contribution from pure Si one can clearly distinguish Ge-Ge, Si-Ge and Si-Si vibrational modes as well as some local Si-Si modes which are well known from SiGe alloys. As about 60% of the surface in each layer is covered by dots we expect 80% of the deposited Ge material to be in the QDs. Hence the remnant spectrum includes mainly contributions from the Ge QDs. The presence of high order folded longitudinal acoustic (FLA) modes in the low frequency region up to 180 cm⁻¹ is a sign for a high quality superlattice with sharp interfaces between dot and spacer layers. The observation of a weak but well defined Si-Si mode at 501 cm⁻¹ originating mainly from the Ge-QDs also indicates sharp interfaces and a rather homogeneous and high Ge content within the QDs. An average Ge content of 80% and lateral strain of −3.4% in the QDs can be obtained from the relative intensities of the Ge-Ge mode to Si-Si mode and the strain induced shift of the Si-Si mode using a biaxial strain model. The validity of this model as well as the deduced composition and strain values have been verified the LO-TO phonon splitting observed from polished facets. The as-grown sample thus shows a small degree of intermixing between Si and Ge and weak (just local) relaxation of the Ge QDs. Annealing of this QD structure (see Fig. 1) leads to a decrease of the intensity of the FLA modes. The simultaneously rising intensity of the Si-Si mode indicates increasing intermixing. This intermixing has been consistently quantified in the same manner as for the as-grown sample. It indicates a low variation of the silicon content and strain in the core regions of the QDs. In Raman studies of modulation p-doped structures at low temperature we observe a distinct Raman signal at 850 cm⁻¹ (105 meV).

Fig. 1. After subtraction of the contribution from pure Si one can clearly distinguish Ge-Ge, Si-Ge and Si-Si vibrational modes as well as some local Si-Si modes which are well known from SiGe alloys. As about 60% of the surface in each layer is covered by dots we expect 80% of the deposited Ge material to be in the QDs. Hence the remnant spectrum includes mainly contributions from the Ge QDs. The presence of high order folded longitudinal acoustic (FLA) modes in the low frequency region up to 180 cm⁻¹ is a sign for a high quality superlattice with sharp interfaces between dot and spacer layers. The observation of a weak but well defined Si-Si mode at 501 cm⁻¹ originating mainly from the Ge-QDs also indicates sharp interfaces and a rather homogeneous and high Ge content within the QDs. An average Ge content of 80% and lateral strain of −3.4% in the QDs can be obtained from the relative intensities of the Ge-Ge mode to Si-Si mode and the strain induced shift of the Si-Si mode using a biaxial strain model. The validity of this model as well as the deduced composition and strain values have been verified the LO-TO phonon splitting observed from polished facets. The as-grown sample thus shows a small degree of intermixing between Si and Ge and weak (just local) relaxation of the Ge QDs. Annealing of this QD structure (see Fig. 1) leads to a decrease of the intensity of the FLA modes. The simultaneously rising intensity of the Si-Si mode indicates increasing intermixing. This intermixing has been consistently quantified in the same manner as for the as-grown sample. It indicates a low variation of the silicon content and strain in the core regions of the QDs. In Raman studies of modulation p-doped structures at low temperature we observe a distinct Raman signal at 850 cm⁻¹ (105 meV).

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It is attributed to electronic Raman transitions from localised heavy (hh) to light hole (lh) states (topmost spectrum in Fig. 2). With increasing annealing temperature the Raman line shifts to lower energy (Fig. 2). Intermixing of Si and Ge reduces the strain-induced hh-lh splitting and the quantum confinement (mainly along growth direction) so that the hh-lh transition energy decreases significantly. Fig. 3 shows the integrated intensity of the Ge-Ge phonon mode and the hh-lh transition depending on the Raman excitation energy. The Ge-Ge mode shows a resonance at 2.3 eV assigned to the E₁ gap. A common resonance of the phonon mode and the electronic transition is observed at 2.5 eV. This Raman resonance is assigned to the direct bandgap E₀ of localised hh and virtually bound Γ–electrons in the small, Ge-rich QDs. The hh-lh Raman transition and the

\[ \text{Raman shift [cm}^{-1}\text{]} \]

\[ 75 \quad 100 \quad 125 \quad 150 \]

\[ 600 \quad 800 \quad 1000 \quad 1200 \quad 1400 \]

\[ 75 \quad 100 \quad 125 \quad 150 \]

Fig. 2: Electronic hh-lh Raman transitions observed from a p-doped as-grown Si/Ge QD structure and structures annealed at varied temperatures.

\[ \text{Raman shift [meV]} \]

\[ 75 \quad 100 \quad 125 \quad 150 \]

\[ 600 \quad 800 \quad 1000 \quad 1200 \quad 1400 \]

\[ 75 \quad 100 \quad 125 \quad 150 \]

Fig. 3: Integrated intensities vs. excitation energy for the Ge-Ge mode and the hh-lh transition.

E₀ resonance of localised states at the direct bandgap represent the first experimental indications of localised electronic transitions within small Si/Ge quantum dots. The detailed structural parameters from AFM and Raman measurements served as an input for valence bandstructure model calculations using the simulator nextnano³ without adjustable parameters. For a lens-shaped, 1.8nm high Ge QD containing 80% Ge the model yields a splitting of 101 meV between hh- and lh-like ground states. This is in very good agreement with the observed Raman shift of 105 meV. Additionally, this non-refined simulation results are also consistent with other experimentally observed transition energies. They predict a E₀ gap of 2.3 eV (compared to the resonance at 2.5 eV), a type-II recombination energy of the hh ground state hh₀₀₀ with Δ-electrons (within Si) of 770 meV and a hh₀₀₀-hh₁₀₀ splitting of 42 meV due to lateral confinement. The latter two are in good agreement with the observed photoluminescence energy of 730 meV of the studied samples and earlier admittance measurements performed on similar Si/Ge QD structures.

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Strain-symmetrized Si/SiGe quantum cascade structures

Ingo Bormann\textsuperscript{1}, Hubert Riedl, Karl Brunner\textsuperscript{2}, and Gerhard Abstreiter

In the long lasting quest for efficient Si based light emitters a promising path seems to be the fabrication of Si/SiGe based intersubband emitters. This approach is based on the quantum cascade scheme, that has successfully been used in lasers for the whole spectral region from mid-infrared to terahertz within the III-V semiconductor systems. Since no interband transitions but only intersubband transitions within the same band are involved, the main obstacle to silicon based light emitters – the indirect band gap – is irrelevant. In Si/SiGe structures the band offset occurs mainly within the valence band. Thus, the quantum cascade must be implemented with hole subbands. Electroluminescence (EL) of first structures grown pseudomorphically on Si substrates has already been demonstrated. However Ge has a 4\% larger lattice constant than Si, which leads to highly strained structures and imposes tight limits on the maximum Ge content and maximum layer thicknesses of such structures. This renders it nearly impossible to design a mid-infrared laser structure with a sufficient number of cascade periods. The solution to this problem is the use of SiGe virtual substrates. This enables to grow strain-symmetrized structures lattice matched to the virtual substrate that have EL wavelengths in the mid-infrared with no restriction on the number of cascade periods.

To explore the feasibility of this concept, we have designed and fabricated by means of molecular beam epitaxy a QC structure on a 28\% Ge virtual substrate. Figure 1 shows the calculated band structure of a SiGe QC structure, with an average Ge content of 34\%. It is nearly strain matched to the relaxed buffer which serves as the virtual substrate. Assuming a relaxation degree of 90\% for the relaxed buffer, the net strain remaining in the cascade is no more than the equivalent of a Ge content of 8\%. Thus we do not expect additional strain relaxation in the cascade at our growth temperature of 480°C.

As a virtual substrate we use a “low temperature” relaxed buffer. With this buffer concept we are

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\textbf{Fig.1:} Calculated band structure of a cascade grown on a 28\% Ge relaxed buffer. Shown are valence band edges and states for heavy (black) and light holes (gray) with an applied electric field of 55kV/cm. The arrow indicates the radiative transition at $\Delta E=164\text{meV}$. Inset shows a EL spectrum at 250 mA driving current.
able to achieve small dislocation densities with relatively thin SiGe layers (here 700nm). Furthermore, it is possible to stack buffer layers with increasing Ge content to achieve Ge-rich buffers. Our buffer consists of a 100 nm Si layer grown at low temperature around 390°C with a growth rate of 0.85 Å/s, that servers as a trap for dislocations and 500nm Si$_{0.72}$Ge$_{0.28}$ grown at 525°C.

Figure 2 is a TEM micrograph of the structure. As can be seen most of the dislocations are trapped within the low temperature Si layer. Nevertheless there are still several threading dislocations, that extend through the entire cascade (20 periods). Their density is on the order of 10$^8$ cm$^{-2}$. The cascade is of good quality, especially with smooth interfaces even in the topmost cascade layers. No influence of the buffer dislocation crosshatch on the cascade growth is visible.

The EL intensity and line width (30 meV) of such samples are comparable to that from structures grown pseudomorphically on Si. This indicates that i) remaining dislocations do not harm the EL intensity or electrical transport ii) the cascade structure is homogenous.

Such a relaxed buffer enables the design of a MIR waveguide with a thick QC active region. Figure 3 shows a proposed low loss waveguide design for a future Si/SiGe QC laser. On the substrate side no additional doping is needed to confine the optical wave, due to the refractive index difference of 0.17 for Si$_{0.72}$Ge$_{0.28}$ compared to pure Si. Calculations predict a confinement factor $\Gamma$=41% and losses $\alpha$=28cm$^{-1}$ for a wavelength of 9µm.

Further work to realize a Si/SiGe QC laser structure with such a waveguide is in progress.

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The EL measurements were done in collaboration with Stefan Schmult and Werner Wegscheider from the University of Regensburg.
Growth and Control of Vertically Correlated InGaAs Quantum Dots for Single Quantum Dot Molecule Spectroscopy

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Due to their discrete electronic structure and ultra-long dephasing times, semiconductor quantum dots (QDs) are often considered to be the solid-state analogue of atomic systems. As such, they are widely regarded as a promising nanostructure for solid-state implementations of quantum information processing, the single exciton transition representing a potential qBIT. In order to explore the potential these systems provide for scalability and ultimately demonstrate quantum conditional logic, a fundamental prerequisite is the controllable coupling of individual qBITs (QDs). As first steps towards this challenging goal, the aims of this project are to study the growth of vertically correlated pairs of QDs, investigate their electronic structure and develop device concepts for controlling the interaction strength within such artificial QD-molecules (QDMs).

In the present report, we present investigations of the growth of QDMs and first experiments to externally control the strength of the interdot coupling using static electric fields. Our QDMs are formed by strain-induced Stranski-Krastanov like self-assembled growth of two layers of nominally In\textsubscript{0.5}Ga\textsubscript{0.5}As QDs, separated by a 7nm GaAs spacer. For such spacer thicknesses, vertical self-alignment of dots occurs due to the preservation of the strain-field. In order to optimize the magnitude of the dot-dot electronic coupling, QDMs with similar electronic properties are required. Furthermore, the growth conditions must be optimized such that the QDMs are accessible to sensitive Si-based detectors (E>1.24eV) and sufficiently low surface density for single molecule spectroscopy. Whilst both the influence of varying the total InGaAs coverage and the In:Ga ratio were investigated, only the former results are presented here. The QD layers were deposited at a

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\begin{figure}[h]
\centering
\includegraphics[width=\textwidth]{Fig1.png}
\caption{Room temperature PL scans along constant composition (blue) direction: Single layer sample shows broad QD emission. Low QD surface density is reached in areas with vanishing PL Double layer samples show additional narrow peak superimposed originating from the upper QD ensemble. Decreased coverage (7ML) in the upper layer leads to a blue shift and better spectral overlap. QD growth breaks down earlier for further decreased coverage (6ML)}
\end{figure}
substrate temperature of 500°C and consisted of 8ML of In$_{0.5}$Ga$_{0.5}$As for the lower QD layer, followed by the 7nm GaAs spacer and 8ML, 7ML and 6ML for three samples respectively. In addition, a single QD layer reference sample. All wafers were grown without substrate rotation in order to vary coverage and In:Ga ratio across the wafer due to the configuration of sources in our MBE system (Fig 1-inset).

For characterization of the samples photoluminescence (PL) spectra were recorded along particular directions on the wafer probing the influence of the InGaAs coverage (Fig.1-blue lines). The single layer reference sample exhibits a QD signal that blue shifts strongly into the E>1.24eV range for decreasing surface density (<10µm$^{-2}$). In comparison, the spectra obtained from the double layer samples contain an additional narrower peak at the low energy side of the spectrum that is attributed to the upper QD ensemble and is more homogeneous due to growth on pre-strained substrate. Suitable transition wavelengths for low surface densities were achieved by reducing the coverage in the upper layer to 7ML. A further decrease leads to a quenching of the PL of the upper QD ensemble and the low energy side of the lower ensemble indicating the presence of material transfer between the QD layers.

In further experiments, QDM samples (8ML in both layers) with 7, 10 and 13nm GaAs spacer thicknesses were embedded in the intrinsic region of an n-i-Schottky junction. These devices enable us to tune the relative energetic position of the carrier states in stacked QDs using a gate bias ($V_B$). Bias dependent low temperature PL spectra are presented in Fig.2 for the three samples in a false color representation. All samples exhibit a main luminescence peak (ES-fig 2) from the upper QD ensemble that dominates over the whole bias region for which PL is observed. Most surprisingly, a satellite peak (RS-Fig 2) is observed for all $V_B$, the amplitude of which resonates close to $V_B$=0.60V. Least square fits of two Gaussian peaks performed at resonance provided a measure for the EP→RP peak separation. Splittings of 26meV, 12meV and 7meV were observed for the 7, 10 and 13nm samples respectively as shown in the right panel of Fig. 2. We tentatively attribute RS to the formation of coupled electron states, a conclusion in very good agreement with recent theoretical predictions.

![Fig. 2: Left: False color representation of the PL spectra as a function of applied bias for the 7nm- 10nm- and 13nm-sample. Right: Spectra recorded at $V_B$ indicated by dashed lines in the false color plot. Dashed and dotted lines are fitted curves.](image-url)
Integration and coupling of quantum dots in microresonators


The principal aims of this project are the investigation and control of the light-matter interaction between fully localised (0D) electronic states, in semiconductor quantum dots (QDs), and quasi 0D-photonic states realised using photonic crystal microresonators. The discrete electronic structure of such QDs are by now well established and have led to a number of proposals for their implementation as non-classical (single photon) light sources as well as numerous applications in quantum information processing. Photon antibunching has recently been demonstrated in the temporal emission characteristics from individual quantum dots, but the realisation of a high efficiency single photon source is crucial for the successful future implementation of both quantum cryptography and photon based quantum information processing architectures. One promising approach to realise this challenging goal is to utilise defect microresonators incorporated into active 2D photonic crystals (Fig. 1) containing quantum dots. Such structures are essentially 2D dielectric Bragg mirrors, which introduce a periodic refractive index contrast within the plane of propagation. This results in the opening of band-gaps for photon propagation in the plane and allowed photon propagation bands (blue lines – fig 2).

Introduction of defects (missing holes) into such crystals disrupts the translational periodicity and results in the formation of highly localised light states within the photonic bandgap, in complete analogy to electronic defects in crystals. An example is shown by the red curves in fig 2, for a hexagonal lattice of circular air holes (a=300nm, r=110nm) with a H2 defect (seven missing holes) demonstrating that typically 10-15 highly localised modes exist within the bandgap.

As for monolithic optical microcavities based on wavelength size cavities with planar Bragg reflectors, photonic crystals provide strong potential to study, modify and control the light-matter interaction in the weak-coupling (irreversible) regime. Therefore it is important to achieve precise control over the energetic position and the size of the photonic bandgap, achieved in practice by varying the lattice period -a and air fill factor (f=ť(r/a)²).

Figure 1: Cross sectional sketch of a patterned active 2D PC waveguide containing a single layer of InGaAs quantum dots

Figure 2: Calculated bandstructure and TE-modes for a H2 defect (seven missing holes) in a 2D hexagonal lattice of circular air holes.

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We have fabricated first 2D photonic crystal microresonators based on an AlGaAs-GaAs slab waveguide structure containing high density (>200µm⁻²) In₀.5Ga₀.5As QDs as depicted schematically in Fig. 1. The waveguide structure consists of the following layers grown using molecular beam epitaxy. Firstly, a 1000nm thick Al₀.35Ga₀.65As (n=3.33 at λ~1nm) waveguide cladding layer is deposited, followed by the 400nm thick Al₀.12Ga₀.88As core region (n=3.46). At the midpoint of the waveguide core, a single layer of In₀.5Ga₀.5As QDs is incorporated. The QD layer was grown without wafer rotation such that a gradient of the dot density is achieved across the wafer between <5µm⁻² to >300µm⁻² for single and many dot experiments. Finally, the waveguide was completed with 10nm GaAs to passivate the surface. After growth, photoluminescence (PL) studies were performed at low temperatures to characterise the emissive properties of the QD layer before patterning with the photonic crystal.

A series of 2D defect photonic crystal microresonators were then defined using electron beam lithography and chemically assisted ion-beam etching. The waveguide was patterned with a hexagonal lattice of air holes (Fig. 3). To enable deep (>1μm) etching of cylindrical air holes with a~100-300nm diameter, shadow masks with high durability are essential. Therefore the first processing step is the deposition of a ~100nm thick Si₃N₄ layer with plasma enhanced chemical vapour deposition (PECVD) using SiH₄/N₂ process gases. In the next step a high-resolution, 100nm thick e-beam photoresist (PMMA 950K) is exposed and developed and the pattern transferred into the Si₃N₄ by dry-etching step using CF₄ in a PECVD-reactor. The Si₃N₄ mask enables us to deep etch cylindrical holes through the AlGaAs waveguide core with an aspect ratio of 1:10 for structures with a periodicity of a few 100nm. An example of the high quality of the resulting structures is presented in Fig. 3.

Fig. 4 shows first spatially resolved photoluminescence measurements performed on the 2D PC defects resonators presented in Fig 3 using confocal microscopy. First evidence for the predicted defect mode-structure (Fig. 2) has been obtained, visible in fig 4 as sharp peaks (2-4meV FWHM) superimposed on the inhomogeneously broadened QD emission. This corresponds to cavity Q-factors in the range 300-450, comparable with state of the art for such structures, and should enable us to observe pronounced Purcell enhancements of the spontaneous emission lifetime. Theoretically we estimate a peak enhancement factor of $F_P \sim 13$, and investigations are currently underway to verify this prediction.
The role of the two-dimensional electron gas in the defect spectroscopy of AlGaN/GaN heterostructures

Martin Hermann¹, Barbara Baur, Martin Stutzmann, and Martin Eickhoff

We have performed defect spectroscopy measurements of AlGaN/GaN heterostructures using photothermal deflection spectroscopy, spectrally resolved photocurrent and persistent photoconductivity. To study the influence of the two-dimensional electron gas (2DEG) formed at the heterointerface, we have varied the density of electrons in the 2DEG by varying the barrier thickness of Al₀.₃Ga₀.₇N/GaN heterostructures from 0 to 35 nm as well as the Al-concentration in AlGaN/GaN heterostructures from 0 to 30% for a constant barrier thickness of 30 nm. In both series of samples, which were grown by plasma assisted molecular beam epitaxy on c-plane sapphire, the formation of the 2DEG causes characteristic changes of the observed spectra:

A systematic increase of sub bandgap absorption in photothermal deflection spectroscopy (PDS) measurements can be observed with increasing barrier thickness and Al-content (c.f. Figure 1). In this case the signal arises from the absorption of the light by transitions of 2DEG-electrons, which cannot be observed in photocurrent measurements as no photocarriers are generated in the channel.

![Figure 1: PDS spectra of Al₀.₃Ga₀.₇N/GaN samples with barrier thicknesses varying from 0 to 25 nm. For photon energies below 3.4 eV an additional absorption due to the AlGaN barrier and the electrons in the 2DEG is observed.](image)

Similar results have been obtained using spectrally resolved photocurrent measurements (Figure 2). A characteristic rise in photocurrent occurs in both series of samples, and is attributed to the formation of the 2DEG and not to an effect of the AlGaN barrier. The rise of photocurrent is caused by an increased “collection efficiency” of photogenerated carriers.

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by the 2DEG, quantified by an enlarged $\mu \tau$-product, where $\tau$ is the recombination lifetime of nonequilibrium carriers.

The enhanced lifetime of photocarriers in the 2DEG of AlGaN/GaN heterostructures is also demonstrated by persistent photocurrent measurements. The decay time of photogenerated carriers was analysed for different excitation energies and as a function of the sheet carrier density in the 2DEG. If a 2DEG is present, we find an increase of the decay time by almost two orders of magnitude at a photon energy around 2.5 eV (c.f. Figure 3), which can be attributed to the optical excitation of electrons out of a defect level in the GaN buffer layer. The onset of this transition at $\approx 2.5$ eV has also been measured by light-induced electron spin resonance in GaN grown by MOCVD. Variation of the 2DEG density shows that only in the case of carriers confined at the interface, these long decay times of the photocurrent can be observed.

**Figure 2:**
Photocurrent spectra of Al$_{0.3}$Ga$_{0.7}$N/GaN samples with barrier thicknesses varying from 0 to 30 nm. The 2DEG varies from 0 to $1.5 \times 10^{13}$ cm$^{-2}$. The sub bandgap photocurrent is strongly enhanced with increasing 2DEG density due to the longer lifetime of electrons in the 2DEG.

**Figure 3:**
Decay time of the persistent photocurrent versus photon energy for AlGaN/GaN heterostructures with different 2DEG densities. The sample with the higher 2DEG density shows a strong increase of the decay time around 2.6 eV. In the inset the decay time versus 2DEG density is shown for a photon energy of 2.8 eV.

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Determination of band offsets in type-II GaPSb/InP heterostructures grown by gas source MBE

Fabian Köhler1, Gerhard Böhm, Ralf Meyer, and Markus-Christian Amann

Type-II superlattices can be applied as modulator zones in tunable twin-guide (TTG-) lasers. The concept of this laser type is described in detail in the article about “widely tunable twin-guide lasers” in this issue. Due to their spatial separation of conduction band minimum and valence band maximum, type-II superlattices cause an enhanced carrier lifetime in case of injection (fig. 1). For the applicability in optical networks, we fabricate TTG-lasers that emit at 1.55 µm. Therefore, the bandgap of the type-II superlattice ought to be larger than 0.8 eV in order to prevent absorption.

In accordance with device technology, type-II superlattices should be fabricated with aluminum-free materials that allow for etching and a later regrowth of the sample. To fulfill these conditions, we have grown GaPSb/InP superlattices and measured the bandgap of GaPSb as well the type-II offset in GaPSb/InP.

We used solid gallium, solid indium, phosphine and triethyl-antimony (TESb) as precursors. Solid group-III sources are used instead of the metalorganic precursors because of the relatively low growth temperatures, which are necessary for growing GaPSb. We used \( T = 540°C \) (thermocouple) instead of \( 590°C \), which is usually applied for metalorganic growth e.g. of InP. At this low temperatures thermal decomposition of metalorganica on the wafer surface is insufficient resulting in a growth rate that is very small and has a strong dependence on temperature. As a consequence, the implementation of solid Ga and In into the CBE-machine, which we now can call a gas-source MBE, was absolutely necessary for growing Sb-based materials. Phosphine as well as TESb are introduced into the growth chamber through a high-temperature injector \( (T = 1050°C) \) and are cracked efficiently into \( P_2 \) and \( Sb_2 \).

In order to find a correlation between precursor supply and material composition, thin InP buffer layers (50 nm) were grown on n-doped InP-substrates followed by bulk layers of GaPSb (thickness between 0.2 µm and 1µm). Ga-flux, phosphine mass flow and substrate temperature were kept constant, and only the TESb mass flow was varied between 0.8 and 2.5 sccm. Fig. 2 shows the correlation between TESb flow and the lattice mismatch of the layer. We found a linear correlation except for very high TESb flows. With this correlation, we were able to grow reproducibly layers which reveal only a small lattice-mismatch and could therefore be grown pseudomorphically with large layer thickness. On these samples we

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measured photoluminescence (PL) in order to get information about the bandgap of GaPSb lattice-matched on InP as well as about the type-II offset in GaPSb/InP heterostructures, as literature does not provide much information about this material. Note that the conduction band edge of GaPSb is expected to be higher than that of InP (staggered configuration). Consequently, the GaPSb-layer must be thick enough, so that not all of the generated electrons relax into the InP. Only in this case, direct transition in the GaPSb and, in consequence, its band gap can be measured. Fig. 3 shows the PL-measurement of a 0.5 µm GaPSb layer at 4.2 K. The high energetic peak is attributed to the direct bandgap of GaPSb, whereas the broader low-energetic peak is due to the spatially indirect transition between the conduction band of InP and the valence band of GaPSb and therefore gives information about the type-II offset. While the peak position of the direct recombination is not significantly affected by the laser power, the type-II transition energy raises with enhancement of the laser power. Therefore, we extrapolated the type-II transition energy to zero laser power in order to calculate the type-II transition energy. We measured a bandgap of 0.901 eV at 4.2 K for GaPSb, which is in accordance with the 300 K bandgap of 0.845 eV found by Shimomura [1]. For the type-II transition we determined a value of 738 meV. The band structure of the GaPSb/InP interface can be specified, as demonstrated in fig. 4, using the well-known bandgap of InP at 4.2 K. We found band discontinuities of 0.163 eV for the conduction band and 0.685 eV for the valence band.

With regard to the application in tunable lasers emitting at 1.55 µm ($E = 0.8$ eV) the type-II transition energy is smaller than the energy of the laser light. This would lead to absorption. Therefore, we are going to increase the bandgap in the GaPSb layer by enhancing the phosphorous fraction of GaPSb from 35% (lattice-matched on InP) towards higher values. To keep the lattice parameter constant, indium is concurrently appended.

In conclusion, we established the growth of lattice-matched GaPSb on InP. By photoluminescence, the bandgap as well as the type-II transition energy in GaPSb/InP heterostructures were measured. In this way, the values of the valence and the conduction band edges of GaPSb can be given on an absolute scale and the band discontinuities at the GaPSb/InP interface are determined.

Type-II superlattices for tunable laser diodes

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Among the different types of tunable laser diodes the tunable twinguide laser diode, c.f. Fig. 1, with its special transversal integration of active and tuning region, is renown for its inherent continuous wavelength control as well as for the large spectral access over 13 nm. Despite significant improvements of technological fabrication with high output efficiency, there is still the major challenge to enlarge the tuning range itself. Exploiting the free-carrier plasma-effect an accumulation of carriers in a forward biased pin tuning diode decreases the refractive index of the device, and, hence, changes the emission wavelength due to the gain-phase condition. In Fig. 2 the refractive index change is schematically illustrated as a function of the tuning current. As can be seen, the negative index change due to the plasma-effect saturates at high currents, mainly, because of the carrier lifetime-carrier density characteristic. Since the carrier lifetime $\tau$ decreases strongly with rising carrier density $N$, at high currents the carriers are no longer accumulated in the tuning region. Further, the enormous dissipation of the carriers by Auger recombination leads to an undesired thermal increase of the device temperature counteracting the refractive index change $\Delta n$ of the former one. In consequence, for a high carrier density and also a large tuning range, a long carrier lifetime is the demand.

Here we report on a type-II superlattice AlInAs-Al$_{0.33}$Ga$_{0.67}$As$_{0.5}$Sb$_{0.5}$ lattice-matched to InP which shows a longer carrier lifetime $\tau$ as well as a larger carrier density $N$ as a func-

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tion of current compared to a common bulk material. The layer sequence of the pin diode (neglecting the n-InP substrate and a p++-InGaAs contact layer) and the corresponding band line up are schematically shown in Fig. 3. As can be seen, whereas the electrons are confined in the AlInAs layer, the holes are spatially separated from them in the AlGaAsSb layer. A particular property of the type-II heterostructure is the reduced effective band gap in respect to the band gap of the single bulk materials. Recombination takes place only due to the weak penetration of the electron and hole wave functions into the corresponding barriers. Therefore, in opposite to the direct recombination in bulk material, the process is called tunneling assisted recombination. Since the recombination of the electron-hole pairs is direct proportional to the square overlap integral of the electron and hole wave functions, a reduction of the recombination of over three orders of magnitude can be expected.

To obtain the lifetime-current characteristic of our type-II superlattice we measured the impedance of the diode as a function of the frequency (100kHz-3GHz) using a network analyzer. The differential carrier lifetime is evaluated by the characteristic time of an equivalent parallel $RC$ circuit. In Fig. 4 the dependence of the differential carrier lifetimes on the nominal current density is shown for the type-II heterostructure and a standard bulk material (InGaAsP, $E_g = 0.95$eV on InP).

![Fig. 4: Lifetime versus current density characteristics](image1)

![Fig. 5: Carrier density versus current density characteristics](image2)

As can be clearly seen, at low current densities the differential carrier lifetime of the type-II heterostructure is over an order of magnitude larger than in the case of the bulk material. Although the carrier lifetime of the type-II heterostructure decreases strongly on current injection due to bandfilling and band bending, it still exceeds the lifetime of the bulk material even at high current densities. It is important to note, that the structure can be further optimized especially using a material combination with higher band offset in the conduction band.

Integrating the measured differential carrier lifetime over the dc bias current, the carrier densities can be calculated as shown in Fig. 5. As a consequence of the larger carrier lifetime in the type-II heterostructure, the carrier density at a given current is enhanced over one order of magnitude in the low current regime as compared to the bulk material. Besides the remarkable lower current consumption necessary to obtain a given carrier density, also a larger maximum carrier density is achieved in the type-II heterostructure. Our results demonstrate the feasibility of type-II superlattices for the application as a tuning region in tunable laser diodes.

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Bias induced desorption of oligonucleotides from Au-electrodes

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The investigation of immobilized oligonucleotides on solid substrates is a lively research area due to the twofold aspects of the topic: for the highly charged nature of DNA, it represents a model system of studying the electrostatic interactions between a polyelectrolyte and a biased electrode surface in electrolyte solution. Moreover, the formidable achievements in the development and commercialisation of DNA-chip technology, as well as its future economic prospects make it an appealing field for industrial research, too.

Here, we report on electro-optical measurements of short (24mer) single-stranded (ss) oligonucleotides immobilized on Au-electrodes. The used DNA samples were chemically synthesised (Oligonucleotide Service Corp. Japan) and modified with a fluorescent label (cyanine dye: Cy3™) at the 5’ end, whereas an alkyl-linker group with a reactive sulphide was attached to the 3’ end (e.g. Cy3-ss24DNA-C6-SH). Au-electrodes of 1mm diameter were lithographically processed onto sapphire substrate wafers and adsorption of DNA onto the Au-surface was accomplished by self-assembly from aqueous solution (24h). DNA adsorption is expected to occur via a covalent Au-S bond, as well as by non-specific physisorption such as electrostatic attraction (image charge), hydrophobic interaction (via the DNA’s aromatic bases), etc.. Measurements in electrolyte solution were conducted in an electro-optical cell, where a three-electrode setup was used to bias the Au-electrodes with respect to a Ag/AgCl reference- and a Pt counter-electrode. A fiber mount (see Fig. 2) was positioned above the electrode to excite and detect photoluminescence (PL) of the Cy3-labelled DNAs. As can be seen from Fig. 1, we observe a rapid increase of Cy3-PL followed by a subsequent decrease upon induction of negative bias potentials to the Au-electrode. This evolution of PL-intensity with time can be explained by a model of three states, taking into account the geometry of the optical measurement, as depicted in Fig. 2:

The PL-emission of Cy3-DNA-strands which are adsorbed on the Au-surface is largely suppressed by efficient non-radiative energy transfer of the photoexcited molecular state to the metal (state ‘1’). Biasing the Au-electrode negatively with respect to the electrolyte solution causes an electrostatic repulsion of the like-charged DNA from the electrode surface, which, at adequate

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Fig. 1: Evolution of the PL-intensity observed at the Cy3-peak-wavelength (circles) over time upon applying a step-like bias sequence (solid line) to the Au-electrode

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electrode potentials, results in desorption of the DNA from the Au-surface. While the released DNA strands are floating freely within the volume of PL-detection (state ‘2’), Cy3-PL is no longer quenched by the metal surface and increased PL-emission can be observed. Finally, Brownian motion forces the released DNA-strands to diffuse out of the optically active volume into a large reservoir (dark state ‘3’). The presented desorption model is supported by additional experimental data (not shown), where (i) no detectable Cy3-emission from the electrode surface is found after applying bias potentials more negative than -1.3V and (ii) spectral shifts in the Cy3 peak emission wavelength allow a differentiation between adsorbed and dissolved DNA. In addition, we developed a simple mathematical rate model according to the preceding discussion and find good agreement with measured data. By varying the electrolytic salt concentration, one can adjust the effective electrostatic screening length, i.e. the Debye length, and hence probe the electrostatic interaction between the biased electrode surface and adsorbed DNA. For this reason we investigated the dependence of the desorption behaviour on the salt concentration upon induction of a single -0.8 V bias step. We find a decreasing desorption efficiency for increasing the salt concentration from 3 to 1600mM (cp. Fig. 3, bottom), which in turn implies screening due to the formation of an electrolytic counter-ion layer between the DNA and the Au-surface. This suggests a conformation of ssDNA on the Au-surface where the hydrophobic bases are pointing towards the metal and the charged, hydrophilic backbone is exhibited to the aqueous solution. A correlated analysis of electrochemical measurements and the PL rise-time however indicates that the desorption process solely occurs during charging of the electrolytic double layer and hence takes place during electrochemical non-equilibrium. This is in accordance with the observation that desorption can be initiated by different bias potentials (cp. Fig.1) which can not be explained by the Gouy-Chapman-Stern theory describing an equilibrium potential gradient. By virtue of the optical measurement geometry, it is also possible to evaluate a setup-specific diffusion time constant, which is characteristic for the mobility of ssDNA in electrolytic solution. Here we observe a decreasing time constant with decreasing screening length (see Fig. 3, top), which is possibly related to the effect of the DNA’s counter-ion cloud thickness on its mobility.

Fig. 3: Bottom: Desorption efficiency denoted by the max. bias-induced PL-intensity, normalized by the initial PL-intensity as a function of Debye length
Top: Measured diffusion time constant of 24mer ssDNA in electrolyte solution for varying Debye lengths
The advantages of amorphous silicon for surface hydrosilylation

Andrea Lehner¹, Georg Steinhoff, Martin Stefan Brandt, Martin Eickhoff, and Martin Stutzmann

Stable, densely packed organic monolayers covalently bonded directly to silicon surfaces currently receive significant interest in the field of biosensor applications, e.g. in biochemistry and biophysics, as they in principle allow the detection and utilization of charge transport across the silicon/organic interface. Preparation of such surfaces can be performed by hydrosilylation of H-terminated silicon with alkenes or alkynes [1]. Crystalline silicon (111) surfaces prepared in this way showed good thermal and chemical stability.

Hydrogenated amorphous silicon (a-Si:H) as an easily producible, large area electronic material is currently used for a variety of different applications, such as displays and solar cells. Organic surface modification of a-Si:H could therefore be an important issue for the fabrication of cheap biosensors in silicon technology. An additional benefit of a-Si:H as far as hydrosilylation is concerned is its reduced sensitivity to oxidation compared to that of H-terminated crystalline silicon surfaces [2]. Surface oxidation reduces the efficiency of hydrosilylation, which becomes obvious from the reaction mechanism: Hydrosilylation is a radical-induced reaction which can be initiated either by free radical initiators, UV light or thermal energy. The reaction starts with a homolytic cleavage of a Si-H surface bond as shown in Fig. 1 (a). The generated silicon dangling bond reacts with the C=C double bond in the alkene to form a Si-C bond (Fig. 1 (b) and (c)). The remaining carbon radical then abstracts a H atom from a neighboring Si-H group and the hydrosilylation reaction proceeds (Fig. 1 (d)).

With the presence of Si-O bonds at the surface the number of reaction centers decreases, because the very stable Si-O bonds cannot be broken by UV light or thermal energies typically used in hydrosilylation reactions. To study the negative influence of surface oxidation on the hydrosilylation in detail, a-Si:H samples were H-terminated and exposed to ambient

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atmosphere for different times (10 seconds, one day, three months). Finally, the samples were hydrosilylated in boiling 1-octadecene for 90 min. X-ray Photoelectron Spectroscopy measurements were performed to obtain the O 1s, C 1s and Si 2p spectra which are shown in Fig. 2.

![Graph showing XPS spectra of hydrosilylated a-Si:H samples for different times before hydrosilylation.](image)

**Fig. 2:** XPS spectra of hydrosilylated a-Si:H samples which have been exposed to ambient atmosphere for different times before hydrosilylation: (———) 10 seconds, (………) one day and (+++) three months. The C 1s peaks are multiplied by a factor of 2. With increasing exposure time, the O 1s and SiO₂ (at 104 eV) peak intensities increase, whereas the Si-Si (at 99.6 eV) and the C 1s peaks decrease. The presence of oxygen at the surface obviously hinders the hydrosilylation reaction.

The sample exposed to air for 10 seconds before hydrosilylation exhibits only a small oxygen peak, which might be caused by adsorbed oxygen or water on the sample surface. The O 1s signal of the second sample is more intense. For a storage time of three months a very large O 1s signal is detected, a clear evidence for the presence of a native oxide on top of the sample, which cannot be altered by the hydrosilylation. The Si 2p peak supports this observation: The Si-Si peak (99.6 eV) decreases with increasing storage time in air. Simultaneously, a SiOₓ peak arises and shifts towards the SiO₂ (104 eV) peak due to the formation of a suboxide and finally of SiO₂. The negative influence of the oxidation of the surface for the hydrosilylation reaction is also indicated by the C 1s spectra. For the shortest exposure time to air a large C 1s peak is detected, an indication of a successful hydrosilylation. This peak strongly decreases for increasing exposure times to air.

a-Si:H is known to be more resistive against oxide formation than crystalline silicon, e.g. we determined the time to form one monolayer of SiO₂ on the surface of a-Si:H in ambient atmosphere to be about 700 h, whereas crystalline silicon oxidizes almost instantaneously under such conditions. Therefore, the use of a-Si:H films for the fabrication of hydrosilylated silicon surfaces could have the additional benefit of allowing the substrate to be treated under ambient conditions during surface functionalization.


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Organically passivated GaAs/AlGaAs heterostructures for biosensor applications


Chemically functionalized semiconductor devices are gaining potential to act as smart substrates for the electric or optical detection of biological or chemical processes. Gateless, laterally patterned planar field effect transistors from surface near two-dimensional electron gases (2DEGs) are promising candidates for such sensor devices due to their superior electrical properties and high sensitivity to even minor changes in surface potential. For applications of these surface functionalized 2DEG devices to biological systems, an interface layer between the semiconductor surface and the organic system above (e.g. living cells, membranes, proteins…) has to accomplish biocompatibility of the substrate, i.e. its electrochemical stability under physiological conditions, and act as an effective signal transducer to the conductive channel.

We prepared a resistor device starting from a MBE-grown GaAs/AlGaAs heterostructure containing a high mobility 2DEG at 60nm below the surface. A Hall-bar structure with an actively measured area of 500 x 500µm was transferred to the sample using standard optical lithography combined with wet chemical mesa etching and alloyed NiGeAu contacts. Reference samples were covered with a standard TiAu Schottky gate. The organic surface modification was performed by immersing freshly prepared and cleaned substrates into 0.1mM mercaptobiphenyl solution in dry ethanol at 50°C for 20h leading to the formation of self-assembled monolayers (SAMs). Physisorbed molecules were removed after the grafting process by several cleaning steps. Some samples were exposed to e-beam irradiation leading to a cross-linking of the phenyl rings to increase the chemical stability.

To evaluate the influence of the organic passivation on the electrical properties of the sample we performed conductance measurements in ambient conditions with a four point measurement geometry on a Hall-bar structure as can be seen in Figure 1. Mercaptobiphenyls with different endgroups (-H, -OH, -CH₃) and thus varying molecular dipole moment were grafted onto the GaAs substrate and the sheet resistance of the samples was measured before and after the monolayer coating. The observed increase in sheet resistance induced by the monolayer deposition, as shown in Figure 2, can be explained understanding the surface functionalized heterostructure as a Field Effect Transistor (FET) where the actual surface potential governs the band bending and therefore the resistance of the de-

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vice. Treating the adsorbed layer of densely packed mercaptobiphenyls with intrinsic molecular dipole moment as a parallel plate capacitor, we estimated the equivalent potential drop across this layer to \( \Delta \Phi_H = 180\, \text{mV}, \Delta \Phi_{OH} = 230\, \text{mV} \) and \( \Delta \Phi_{CH3} = 320\, \text{mV} \). To achieve the same relative resistance changes in reference measurements using a standard Schottky gate directly controlling the surface potential a voltage of \( \Delta \Phi_{\text{gate}} = 90\, \text{mV} \), 120 mV and 140mV was applied, respectively. The order of magnitude of the corresponding voltages fits quite well.

To demonstrate the possible potential of the mercaptobiphenyl passivated device to act as a biosensor we conducted stability measurements in aqueous solutions in an experimental setup as shown in Figure 1. The sample resistances of a bare and a mercaptobiphenyl passivated device immersed in a phosphate buffered saline solution were measured in a two point geometry. Here, the monolayer coated sample was cross-linked by low energy e-beam irradiation in order to increase the chemical stability. As can be seen in Figure 3, the bare sample degraded rapidly whereas the passivated device showed quite good stability over at least three hours.

**Fig. 2:** Mean sheet resistance vs. projected molecular dipole moments for surface functionalized heterostructures passivated by mercaptobiphenyls with different endgroups.

**Fig. 3:** Sample resistance vs. immersion time for a surface functionalized GaAs/AlGaAs device in 0.1M phosphate buffered saline (PBS) solution at pH 6.50, ionic strength \( I = 0.1\, \text{M} \) measured at \( T=23^\circ\text{C} \), compared to an un-functionalized reference sample. The bare sample (■) degrades rapidly while the mercaptobiphenyl passivated and cross-linked sample (●) remains comparably stable.

In conclusion, we demonstrated the influence of self assembled monolayers of mercaptobiphenyls with varying dipole moment on the electrical conductance of shallow 2DEGs and thus showed the high sensitivity of the devices to surface potential changes. Together with the confirmed increased stability of the mercaptobiphenyl passivated samples in aqueous physiological buffer solutions this demonstrates the device’s potential as a sensor for biological and chemical processes.

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Oxidation of gas-sensitive GaN and AlGaN Schottky diodes

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Published data on the electrical properties of GaN and AlGaN Schottky diodes such as barrier height and ideality factor show a large scatter. In addition, threading dislocations as the most common structural defects are known to cause a high reverse current and thereby substantially lower the device performance.

One way to overcome these limitations is the passivation of leakage current paths prior to the deposition of the Schottky metal. For this purpose we have studied the influence of surface oxidation on the device performance and compared the electrical characteristics of Pt:GaN Schottky diodes on as deposited GaN to those prepared on thermally oxidized samples. The oxidation was carried out in dry or wet atmosphere at temperatures of 600°C, 700°C and 800°C, respectively. As shown in Fig.1, oxidation at 700°C results in a substantial improvement of the device performance. The reverse current is reduced by more than three orders of magnitude and the effective barrier height increased from 0.5eV for the non-oxidized sample to 0.9eV and 0.8eV for the wet and dry oxidized diodes, whereas series resistance the remained almost constant.

The surface morphology after the oxidation process at 700°C was analyzed by atomic force microscopy. As shown in Fig.2, the oxidation process preferentially starts at small islands, which probably mark the positions of dislocations at the surface, whereas the rest of the surface is covered by a thinner oxide layer, explaining the only small increase the series resistance. As a promising technique for selective passivation of leakage paths which cause an increase of the reverse current, we also have investigated anodic oxidation in 5% KOH. We have observed a

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decrease of the reverse current by three orders of magnitude after anodic oxidation. Thus, both oxidation processes lead to a significant device improvement. The influence of surface oxidation was also investigated by in-situ prepared Pd:GaN Schottky diodes, which exhibit an oxide-free metal-semiconductor interface. Comparison of the electric characteristics to those of conventional ex-situ deposited diodes confirms the device improvement by oxidation. In this case already the native oxidation in atmosphere was sufficient. Figure 3 shows a significantly higher reverse current and lower Schottky barrier for the in-situ diode.

The sensitivity towards hydrogen of the in-situ and ex-situ deposited diodes was also studied. The ex-situ diode showed a large voltage shift upon exposure to hydrogen (Figure 4), which is reversible by subsequent removal of the H\textsubscript{2} gas. In contrast, in-situ prepared diodes exhibited almost no voltage shift. This leads to the conclusion that the metal oxide at the metal semiconductor interface rather than the clean GaN surface provides the necessary adsorption sites for hydrogen.

**Fig.3:**
IV-characteristics of Pt:GaN Schottky diodes with in-situ and ex-situ deposited Schottky contacts. In-situ deposition results in a higher reverse current and a lower effective barrier height.

**Fig.4:**
Hydrogen sensitivity of an ex-situ deposited Pd:GaN Schottky diode. Exposure to hydrogen leads to a large shift in the IV-curve. In contrast, in-situ deposition of the Pd contact results in a very small hydrogen sensitivity.
**pH-Response of GaN surfaces and its application for pH-sensitive field effect transistors**

Georg Steinhoff, Martin Hermann, Martin Eickhoff, and Martin Stutzmann

Since the first report of an ion sensitive field effect transistor (ISFET) based on a Si n-channel transistor with a SiO₂ gate layer by Bergveld, there has been great interest in ISFETs for physiological measurements and detectors for biochemical processes. Transistor structures based on GaN or AlGaN/GaN heterostructures grown on sapphire substrates can easily provide metal oxide gate layers, which are known to show high pH sensitivity, and benefit from the high chemical stability the biocompatibility of the III-nitrides. Due to the optical transparency of both, the thin semiconducting film and the sapphire substrate, GaN-based ISFETs allow simultaneous electronic and microscopic detection of processes on the device surface in biophysical applications. Although some work on III-nitride based chemical sensors has been done recently, the ion-sensitivity of GaN surfaces in aqueous solutions has not been systematically investigated so far.

We have studied the electrical response of as-deposited and thermally oxidized GaN surfaces to variations of the H⁺-concentration in electrolyte solutions. As a readout device for the ion-induced changes of the surface potential we have used PIMBE grown transistor structures based on GaN:Si/GaN:Mg double layers and AlGaN/GaN heterostructures shown in Figure 1. Samples A and B consist of a 60 nm thick Si-doped GaN channel (n=3⋅10¹⁸cm⁻³) grown on a 1.5 µm thick GaN buffer layer (n=5⋅10¹⁵cm⁻³) with N-face polarity. The latter was partially compensated with Mg to achieve a high resistivity compared to the thin Si-doped channel close to the surface. Sample B was thermally oxidized in dry oxygen at 700 °C for 2 h, leading to the formation of a thin GaₓOᵧ layer, as observed by XPS measurements. XPS analysis of as-deposited GaN surfaces via the surface sensitive Ga₂p and the O₁s core level spectra also revealed the almost immediate formation of a thin surface oxide after exposure to atmosphere. The AlGaN barrier of the HEMT-structure (sample C) had an Al-content of 28 % and a thickness of 35 nm. The 2DEG carrier density was determined to n₂DEG = 6⋅10¹²cm⁻² by capacitance-voltage measurements. Deposition of an additional 3nm GaN cap layer on top of the AlGaN barrier allowed the comparison of similar surfaces on different readout devices (samples A and C), whereas the sensitivity of different surfaces on equal device structures was analyzed by comparison of samples A and B.

**Fig.1:** Schematic layout of the investigated transistor structures. All samples were grown by PIMBE. The surface of sample B was thermally oxidized at 700°C for 2h.

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The chemical response of the gate surface to changes in the electrolyte composition was measured by adjusting the gate potential $V_G$ via a Ag/AgCl reference electrode and a potentiostat in such a way, that the ion-induced changes in the channel current $I_{DS}$ at constant $V_{DS}$ were compensated. Under these measurement conditions, the pH-dependent change in the gate voltage, $\Delta V_G$, is a direct measure for the potential change at the GaN/electrolyte-interface. A linear behavior over the entire investigated range from pH 2 to pH 12 is observed for all three samples (Fig. 2), with sensitivities close to the Nernstian pH response of 58.7 mV/pH at 23 °C. Identical surface layers on different device structures (samples A and C) lead to almost identical sensitivities, demonstrating that the pH-response is a pure surface effect. As surfaces with native oxide and thermally oxidized samples showed almost no difference in sensitivity, we assume that a Ga$_x$O$_y$ surface layer is responsible for the observed behavior in both cases.

To investigate the transient behavior of the GaN-based ISFET devices, we performed time resolved measurements of the drain source current for $V_{DS} = 250$ mV and $V_G= 0$ mV. The pH of the electrolyte was changed in steps between 0.1 pH and 0.3 pH by titration with diluted NaOH every 30 s. In contrast to reported delay times of some seconds for SiO$_2$ gate ISFETs, all investigated GaN-based devices showed an immediate response to changes in the pH. We estimate a resolution of 0.05 pH for the HEMT-structure and of 0.1 pH for the other two devices. However, further optimization of important device parameters is likely to still improve the device resolution.

**Fig. 2:** Variation of the GaN/electrolyte- and Ga$_x$O$_y$/electrolyte-interface potential as a function of pH. For clarity the datasets of B and C are shifted by a constant potential offset of +100 mV and -100 mV. The electrolyte was a 100mM NaCl, 10mM Hepes solution, titrated with diluted NaOH and HCl for pH adjustment.

**Fig. 3:** Transient behavior of the source-drain current $I_{DS}$ during changes of the electrolyte pH by titration with NaOH. $V_{DS}$ and $V_G$ were kept constant at $V_{DS}= 250$ mV and $V_G= 0$mV. Numbers correspond to pH values measured with a calibrated pH-meter.

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Diamond-based ion sensitive field effect transistors
Jose A. Garrido1, Stefan Kuch, and Martin Stutzmann

Due to its large stability against many chemical agents and to its biocompatibility, diamond ISFETs are a very attractive material in the field of chemical sensing and biosensing. Degradation does not occur while the diamond surface is in contact with aqueous solutions whose electrochemical potentials fall within the large band gap of diamond. Therefore, chemical sensors based on diamond would not require electrochemical passivation of the surface, as in Si-based sensors. The realization of the first prototypes of diamond sensors for gas and glucose-sensing applications has already been demonstrated. Moreover, covalent immobilisation of DNA on diamond using photochemistry has recently been achieved, which confirms the potential usefulness of diamond as a solid substrate for biological molecules. However, in order to complete the promising perspectives which diamond might open in the field of bioelectronics, one piece is still missing: the development of Ion Sensitive Field Effect Transistors (ISFETs) which could be used for electronic detection of biological/chemical events.

Among the different characteristics that diamond has, surface conductivity based on hydrogenated diamond surfaces might be the best option for the fabrication of ISFETs. One of the main reasons is the existence of a quasi-two dimensional hole gas at the surface of hydrogen-terminated diamond, which opens new schemes for electronic detection in biosensing. Here we report on the fabrication and characterization of ISFETs based on hydrogen-terminated diamond surfaces.

Figure 1a shows a schematic view of the fabricated diamond-ISFET. Hydrogenated polycrystalline CVD diamond samples were used in this study, with typical hole mobilities in the 10-50 cm²V⁻¹s⁻¹ range, and a carrier sheet concentration about 10¹³ cm⁻². 100nm-thick Au pads were deposited as drain and source contacts using e-beam evaporation. Drain and source Au contacts were covered with a silicone-glue in order to protect them from the electrolyte solution. The gate area exposed to the electrolyte was typically 1mm in length and 4mm in width. An important characteristic, which makes these devices different from normal ISFETs, is the absence of an oxide or insulator layer on top of the semiconductor. ISFETs were immersed in NaCl and KCl aqueous solutions and operated using a 3-electrode configuration (with a Pt counter electrode and a Ag/AgCl reference electrode). Figure 1b shows the current-voltage curves as a function of the gate reference potential for a diamond-ISFET immersed in a 10 mM NaCl solution (pH = 5.5).

Hydrochloric (HCl) and sulphuric acids (H₂SO₄) were used to change the pH of the solution. The pH sensitivity of the devices was measured by fixing the drain-source current (I_DS = -25μA) and the drain-source voltage (U_DS = -100mV). Changes in I_DS due to changes evaporation. Drain and source Au contacts were covered with a silicone-glue in order to protect them from the electrolyte solution. The gate area exposed to the electrolyte solution was typically 1mm in length and 4mm in width. An important characteristic, which makes these devices different from normal ISFETs, is the absence of an oxide or insulator layer on top of the semiconductor. ISFETs were immersed in NaCl and KCl aqueous solutions and operated using a 3-electrode configuration (with a Pt counter electrode and a Ag/AgCl reference electrode). Figure 1b shows the current-voltage curves as a function of the gate reference potential for a diamond-ISFET immersed in a 10 mM NaCl solution (pH = 5.5).

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in pH were compensated by adjusting in the gate voltage ($U_G$). Figure 2 shows the pH sensitivity curves for H-terminated diamond ISFETs with different surface treatments. Fully hydrogenated diamond surfaces do not show any pH sensitivity (solid squares in Fig. 2), but after a weak oxidation of the surface by ozone, diamond ISFETs show an intriguing pH sensitivity. Increasing the oxidation time results in an increase of the pH sensitivity (in Fig. 2 oxidation-1 is a less effective process than oxidation-2). A pH sensitivity of 65 mV/pH, higher than the Nernst limit (58.2 mV/pH), was obtained in the case of oxidation-2. Even more surprising is the sign of the pH sensitivity. In Fig. 2, the dashed line represents the theoretical pH sensitivity predicted by the site recombination model. Such a model applies to normal ISFETs based on electrolyte/insulator/semiconductor interfaces and assumes that a change in the pH alters the potential drop at the electrolyte/insulator interface. However, in our case this electrolyte/insulator interface does not exist and the site binding theory cannot be applied.

In order to obtain a deeper understanding of the pH sensitivity of diamond-ISFETs, other experiments were carried out. The dependence of the pH sensitivity on the gate potential has been investigated (see Fig. 3). Sensitivity curves were recorded at different initial gate voltages ($U_G$= -0.6 V, -0.45 V and -0.3 V). Decreasing the gate voltage (from -0.6 V to -0.3 V) produces an increase in the pH sensitivity (from 65 mV/pH to 87 mV/pH). This behaviour is not predicted by any established ISFET theory and is still under investigation.

A model based on the Coulomb interaction between ions in the solution and holes accumulated at the hydrogenated diamond surface is being developed. Fully hydrogenated diamond surfaces do not provide any site for charge adsorption, and therefore are not ion sensitive, whereas oxygen sites at the diamond surface enhance the adsorption of $H^+$, which produces a depletion of the hole channel (the lower the pH the higher the proton concentration, and the stronger the hole depletion).

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**Fig. 2:** $U_G$ vs. pH characteristic for $I_{DS}=-25 \mu A$.

**Fig. 3:** $U_G$ vs. pH as a function of the ISFET bias point.
Development of a biosensing device based on SOI


Efficient detection of specific or unspecific interactions at solid/liquid interfaces is becoming increasingly important for a wide variety of applications and technologies, ranging from industrial and biomedical applications to basic research. The most prominent need for applications is the label free detection of biomolecular interactions as the common fluorescent labelling techniques are limiting the applicability of DNA or protein arrays. Increasingly sophisticated techniques are being developed to provide highly sensitive and specific detection of molecules bearing the potential of simultaneous parallel detection. For the electrical detection of molecular interactions physical principals of semiconductor field effect transistors can be used: variations of the surface potential induce changes of surface band bending and thus of the conductivity of the sensing layer, as e.g. realized in the extensively studied technique of Ion Sensitive Field Effect Transistors (ISFETs).

Here we present a device based on Silicon-on-Insulator (SOI) technology, enabling the detection of changes of electrolyte concentrations and variations of small number of charged biomolecules, evoked by changes of the surface potential. In the SOI substrates the conducting layer is limited to a thin surface-near Si sheet whose conductivity is strongly dependent on variations of the surface potential. We utilize a four point resistance measurement in a Hallbar geometry to determine the lateral resistance of the sensing layer. A variation of the salt concentration of the electrolyte can be detected over 5 orders of magnitudes in excellent agreement with the Poisson-Boltzmann theory. The unspecific adsorption of Poly-L-lysine was detectable already at concentrations of 1nM (8 ng/ml). The measurements show that variations of surface charges of 0.025 e⁻ per nm² or one electronic charge per 40nm² can be detected by the device.

Fig.1: Optical micrograph of the SOI device. Bright regions indicate mesa-etched conductive Si (30-80 nm) on top of buried SiO₂ (200 nm). Inset: Complete chip with outer metal contacts. In electrolyte, a flow chamber limits the contact of liquids to the inner Hallbar.

In order to relate the sheet resistance to the potential at the oxide surface, calibration measurements in buffer solution with an Ag/AgCl reference electrode were carried out. The surface potential and the voltage of the reference electrode add linearly to the total applied voltage at the oxide/electrolyte interface. Thus, the relation between the square resistance and the potential of the reference electrode is the same as the relation between surface potential and square resistance, except for a constant offset. Additionally, the dependence of the sheet resistance from an applied back gate voltage $V_{bg}$ was determined, yielding the

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working point of the device. The back gate voltage with the highest $dR_{\text{square}}/dV_{bg}$ was chosen to get the highest possible sensitivity.

The surface potential of a solid substrate in contact with an electrolyte solution is well studied and best described by the Grahame equation. The charge density of the silicon oxide surface depends primarily on the pH of the electrolyte solution. The potential at a surface with a constant charge density is controlled by the salt concentration of the electrolyte solution.

![Grahame equation graph](image1)

**Fig. 2:** Left: Square resistance as a function of KCl concentration (solid symbols) compared to theory (Grahame equation, line). Right: Effect of different concentrations of poly-L-lysine in 500mM PBS-buffered KCl solution.

As can be seen in Fig. 2 (left), changes of the electrolyte concentrations were easily detected by the SOI device. The surface charge density of the device was kept constant by controlling the pH ($pH=6.4$) of the solution tightly. In the example shown, we used phosphate buffered (10mM buffer salts) KCl solutions. The Grahame equation for the dependence of the surface potential on the concentration of 1:1 electrolyte, such as KCl in presence of an additional amount of divalent ions and the surface charge density $\sigma$ can be used to fit the obtained data. The theoretical curve corresponds to a surface charge density of $0.02 \, \text{C/m}^2$ and a dissociation of the divalent buffer ions of 20%, both in good agreement with the literature values.

The sensors capability to detect variations of the surface potential due to variations of the surface charge density was studied by adding increasing amounts of positively charged polypeptide poly-L-lysine to the buffer solution. In order to test the sensitivity of the sensor in physiological buffers an electrolyte concentration of 500mM was used (10 mMolar PBS plus 490 mMolar KCl). Fig. 2 (right) shows the change in resistivity after exposure to different poly-L-lysine concentrations. Injection of a solution of 0.1 nM Poly-L-lysine into the fluidic chamber was not detectable, whereas a solution of 1nM Poly-L-lysine (8ng/ml) resulted in a clear change of the conductance. The calibration measurement can be used to relate the observed jump in resistance of $\Delta R = 100\Omega$ to a change of the surface potential of $\Delta \psi = 2mV$. This relates to a charge density of $0.025 \, \text{e}^+ / \text{nm}^2$, neglecting screening effects in the electrolyte solution. This means that the SOI device would be able to detect approximately between 0.1 and 0.5 net charges per biomolecule in physiological solutions.

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GaInAs-based waveguides for high-performance InP-based quantum cascade lasers

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The quantum cascade (QC) laser is a new type of laser that has become a promising alternative to common diode lasers in the mid-infrared spectral region for environmental, medical and industrial applications. High peak output power levels in the Watt range (>3W) and high temperature operation in pulsed and in continuous wave operation have been reached.

In the design of high-quality QC-laser devices, a crucial issue is the reduction of the waveguide losses which are predominantly controlled by free-carrier absorption. Therefore, we focused on design and realization of low loss QC lasers based on plasmon enhanced GaInAs-waveguides. The idea is to take advantage of the abnormality of the refractive index in highly-doped semiconductor materials experienced by wavelengths just above the plasma frequency. The resulting strong decrease of the refractive index enhances the confinement factor of the mode. However, along with the reduction of the refractive index goes a strong increase in the absorption of highly doped layers. Therefore, to reduce the optical losses, the mode penetration in these layers is reduced by sandwiching the active region between thick low-doped waveguide layers. In order to efficiently exploit the plasma effect high accuracy (~±10\%) of the doping level must be accomplished. For a wavelength of about 5.5 $\mu$m, low waveguide losses and high confinement of the optical mode can be achieved by using differently doped GaInAs layers. In the (AlGaIn)As-material system offered by the MBE, GaInAs lattice-matched to InP-substrate shows the widest range of doping levels (Si-doped) with relatively high values of the electron mobility.

Two different laser waveguide structures have been designed and compared, which are referred to as sample A and B (Fig. 1). The active region is the same for both structures, but the waveguide design of sample B has been improved by reducing the losses. Fig. 1 shows the TM-modes and the refractive index profiles. For both structures the plasmon waveguide layer plays a crucial role in suppressing the coupling between the optical mode and the high-loss plasmon mode propagating along the metal-semiconductor interface.

\textbf{Fig. 1:} TM-modes and refractive index profiles of the samples A and B. The calculated optical losses at room-temperature are also indicated. The calculated confinement factor in the active region is for both samples almost the same (~47%).

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The doping level has been chosen high enough so that the plasma frequency (equivalent to a wavelength of 6.4 µm) approaches but does not exceed that of the waveguide mode (5.5 µm). The resulting free carrier absorption losses in this layer are 2500 cm\(^{-1}\). The strong decrease of the refractive index down to 1.95 enhances the confinement factor of the mode and suppresses the mode penetration in the contact region, thus reducing the waveguide losses.

A plot of the threshold current densities at various heat sink temperatures for 2.5 mm long and 30 µm wide representative devices fabricated from the two samples is shown in Fig. 2. The reduction of the threshold current density of about 25 % is in good agreement with the calculated value of 20 %. For such structures, operating temperatures as high as 450 K have been observed. The combination of the low-loss waveguide design together with the high injection efficiency of the injection miniband is the key point for the high-temperature operation of these lasers.

A plot of the emission wavelength at various heat sink temperatures is also shown in Fig. 2. Wavelength and tuning coefficient (~0.9 nm/K) are quite similar for both samples. This confirms the high control and the high reproducibility achieved with the MBE growth. Finally, Fig. 3 shows the threshold current densities - for similar devices with a reduced number of periods - with varying lengths in order to extrapolate the waveguide losses. The observed temperature-dependent waveguide losses could be explained by means of the measured temperature-dependent mobility values in the Drude model, which give losses of \(\alpha_w = 6.5\) cm\(^{-1}\) (300 K) and 2.8 cm\(^{-1}\) (77 K) in very good agreement with the measured values.

In conclusion, we have fabricated high-performance InP-based QC lasers using only GaInAs as cladding material, avoiding heterobarriers between metal contact and active region and making use of an MBE system without phosphor in the fabrication procedure.

Fig. 2: Threshold current density (left) and emission wavelengths (right) vs. heat sink temperature in pulsed mode.

Fig. 2: Threshold current density vs. the reciprocal cavity length at two different operating temperatures for the determination of the waveguide losses.
Index-guiding in InP-based laser diodes by oxidized AlAsSb layers

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We report on thin selectively oxidized AlAsSb layers that provide current confinement as well as optical confinement in InP-based edge emitting laser diodes. The AlAsSb layers are $p$-doped and enable an overall low series resistance of the device of $2\,\Omega$. To investigate the index-guiding of the oxidized region, far-field measurements were carried out. Good agreement between the experimental and simulated far-field patterns indicates that the partially oxidized AlAsSb layers act as a low index-cladding by the native-oxide.

By selective wet oxidation of high Al-content compounds, current confinement and index-guiding can be simultaneously accomplished in III-V optoelectronic devices. Whereas the oxidation of AlAs is a mature technique for GaAs-based devices, the oxidation of AlAsSb is a promising candidate to adapt this technology for InP-based devices. Although AlAsSb exhibits high oxidation rates at low temperatures, upon oxidation of thick AlAsSb layers an interfacial metallic layer is formed by the segregation of Sb besides the aluminum oxide. This interfacial layer causes not only a swelling of the oxidized region and, hence, a deforming of the upper cladding layers, but also, due to the metallic conductivity of Sb, it is not clear, whether the oxidized region could serve as a current aperture as well as a low index cladding in laser diodes. In this paper we report on thin $p$-doped AlAsSb layers which were selectively oxidized to fulfill the demands for a current and index aperture in laser diodes.

A laser structure was employed which consists of three strained InGaAs quantum wells emitting at an emission wavelength of $\lambda = 1.79\mu m$. A 25nm thick AlAsSb layer $p$-doped with Be ($p = 5 \times 10^{17} \text{cm}^{-3}$) was grown 100nm above the active region. After wet chemical etching of 90$\mu m$ broad stripe mesas, the oxidation was carried out in a hot steam atmosphere.

![Fig.1: Schematic cross section of a laser with an oxidized AlAsSb current aperture and the resulting effective index profile in the lateral direction](image)

Due to the high selectivity of the wet oxidation process with respect to the Al-content, only the AlAsSb layers are oxidized, whereas all other layers remain unoxidized. Fig. 1 shows the schematic drawing of such an oxide-confined stripe laser. The active region is defined by the width of the remaining unoxidized AlAsSb layer. The refractive index of the oxidized AlAsSb layer is reduced to a value of about 1.6 for Al$_x$O$_y$, causing a step in the effective refractive index in the lateral direction.

After oxidation, lasers were cleaved and characterized by optical and scanning electron microscopy. Also the light output power-current and the voltage-current characteristics were measured and compared to unoxidized broad-area laser diodes. Furthermore, far-field measurements were carried out and compared to numerical simulations based on the transfer-matrix method.

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Fig. 2a shows an optical microscope image of the sample surface after oxidation at 360°C for 10 minutes. In this laser structure, a 25nm thick AlAsSb oxidation layer is embedded between InAlAs layers. As can be clearly seen, the surface is smooth and shows no deformations. The oxidation depth is uniform and isotropic for the (110) and (-110) directions. Also scanning electron microscope images of the cross section of the oxide show no interfacial layer around the native oxide (c.f. Fig. 2b).

![Fig. 2: a) Optical microscope image of the surface and b) cross sectional SEM image of the InAlAs/AlAsSb/InAlAs stack after partial lateral oxidation](image)

**Fig. 2**: a) CW output power-current characteristic of a 1mm long laser with 5µm current aperture at 20°C; b) measured (squares) and simulated (line) horizontal far-field pattern

Fig. 2a shows a typical continuous-wave (CW) output power-current characteristic of an oxide-confined laser at 20°C. The laser has a cavity length of 1mm and a current aperture of 5µm. The lasing threshold is reached at a current of 66mA. Taking current spreading by carrier diffusion into account, this corresponds to a threshold current density of 940mA/cm². Due to the broad p-contact area the laser shows a series resistance of only 2Ω. The value of the external differential quantum efficiency is about 60% indicating that only low optical losses are present. The far-field pattern at a current of 100mA is depicted in Fig. 2b. The measured full-width at half-maximum (FWHM) angle \( \theta_{FWHM} \) is about 9.5° and agrees well with the calculated FWHM angle of 9.4°.

We have demonstrated the capability of thin oxidized AlAsSb layers for achieving lateral index-guiding in edge emitting lasers on InP substrates. Besides providing index guiding, the oxidized AlAsSb layers also act as an electrical insulator and give excellent current confinement. Our results encourage the development of vertical-cavity surface-emitting laser diodes that employ native-oxide AlAsSb layers for the current confinement and index-guiding.

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Buried heterostructure lasers based on chemical beam epitaxy

Thomas Jacke, F. Köhler, René Todt, Ralf Meyer, and Markus-Christian Amann

Although several progress has been achieved with other laser designs, in the field of optical data transmission, buried heterostructure (BH) lasers are still very attractive light sources providing high output power and nearly Gaussian beam shape. Because of the strong lateral wave-guiding of the embedded mesa (high lateral and transversal variation of the effective refractive index) this concept is ideal for tunable single-mode lasers. For such devices at least 3 epitaxy steps are needed. The appropriate surface treatment before regrowth has been identified as a very critical step in the fabrication of BH-lasers and had to be studied extensively.

On misoriented InP substrates multiple quantum well structures in the InGaAsP material system were grown by chemical beam epitaxy (CBE). The mesas were formed by etching the Si$_3$N$_4$-patterned wafer with an ECR-RIE process. Afterwards, the mesas were embedded laterally in CBE with $n$-type InP by selective area growth (no material deposition on top of the Si$_3$N$_4$ capped mesa). After removing the Si$_3$N$_4$-mask, the upper $p$-cladding and the contact layers were grown.

One complementary sample has been grown with $p$-type InP at the second embedding overgrowth step, so that the interface between second and third epitaxy is uniquely $p$- on $p$-type. Electrically, there should be no difference, since the current path in the low voltage regime (below the bandgap of InP) is dominated by the low bandgap material ($pin$-diode). In electro-optical measurements the sample with $p$-type InP in the second epitaxy shows laser activity (extrapolated threshold current density of 0.5kAcm$^{-2}$), but an irregular $IV$-curve (high ideality factor). On the contrary, the $n$-type second epitaxy samples exhibit only an $IV$-characteristic of a regular InP $pn$-diode and no lasing action. Further measurements confirmed that there is an interfacial blocking layer between two $p$-type InP layers (s. Fig. 2) but no blocking layer at $nn$, $pn$- or $np$-InP interfaces. This prevents the current from passing through the low bandgap pin-diode path. Accordingly, the surface of the sample must be contaminated with $donator$ atoms (carbon, silicon or sulfur) before overgrowth in CBE. This is in good agreement with $CV$-measurements that show a deep “doping gap” around the epitaxial $pp$-interface.

The first step was to identify and possibly eliminate the contamination source. However, different treatments of the samples revealed, that simply the exposure to

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$\text{Fig. 1: }$ SEM of a Buried Heterostructure Laser

$\text{Fig. 2: }$ Cross section of a BH-Laser with blocking layer (red cross-marks) between $p$-doped InP of first and third epitaxy

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cleanroom air for about 1h was sufficient to build up an interfacial \( n \)-layer. The comparison by means of SIMS between a sample with 2min exposure to air (no blocking effect) and another with 2h exposure to air, proved carbon to be the undesired doping element at the interface (Fig. 3): The carbon signal of the 2h-sample showed a peak one order of magnitude higher than the cross-check sample, whereas the oxygen signal is just increased by a factor of two.

The \( n \)-type character of the contamination suggests, to eliminate the \( n \)-layer through compensation with high \( p \)-doping. In CBE, a maximum incorporated doping concentration of beryllium of about \( 2 \cdot 10^{18} \text{ cm}^{-3} \) can be achieved, which turned out to be insufficient to compensate for the contamination. A zinc diffusion process with much higher doping level led to a reduction of the blocking effect, but the regrowth morphology was very poor.

Several wet etching and solvent processes have been tested to remove the contamination together with the uppermost semiconductor atomic layers. Despite of the different etching mechanisms, no such treatment had an effect on the blocking layer. Thus, carbon containing molecules remain probably adsorbed on the semiconductor surface getting incorporated later during the deoxidation process in the CBE-chamber.

Also different heating procedures in CBE have been tested. The heating temperature has been increased up to \( 600^\circ \text{C} \) and the deoxidation duration has been extended up to one hour without any noticeable result on the contamination. Another possibility to avoid the blocking structure, is, to modify the surface composition with an arsenic stabilization before regrowth. Arsenic atoms replace phosphorous atoms and possibly carbon-phosphorous bonds. So far, no working arsenic stabilization procedure has been found in CBE (surface damage).

A very successful way to remove carbon contamination from the sample surface, turned out to be an exposure to ultra-violet light in oxygen atmosphere. Under the radiation of the mercury-lamp \( \text{O}_2 \) is converted into \( \text{O}_3 \), which reacts with the adsorbed organic molecules producing \( \text{CO}_2 \) and \( \text{H}_2\text{O} \). Furthermore, the InP surface gets passivated by ozone preventing further contamination. The oxidized surface could be removed in CBE with a longer heating period. \( IV \)-curves of such samples showed perfect conductivity. In \( CV \)-measurements the doping depletion at the interface was even less than the cross-check samples with 2min air exposure.

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**Fig 3:** SIMS of a sample with two hours air exposure (colored lines) and the cross-check with 2min air exposure (black lines). Upper chart: oxygen, lower chart: carbon
GaSb-based low threshold 2.7 µm laser

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Today there is a growing interest in laser sources for the mid-infrared-(MIR) wavelength range above 2.5 µm. Diode lasers emitting in this wavelength range are attractive devices for molecular spectroscopy, trace gas sensing and chemical process control. Up to a wavelength of 2.5 µm quantum-well (QW) lasers based on the GaInAsSb/AlGaAsSb material system have shown excellent performance regarding threshold current density and characteristic temperature. However above this wavelength the performance decreases significantly.

In our group we fabricated GaInAsSb/AlGaAsSb multiple quantum well lasers with an emission wavelength of 2.7 µm with low threshold current densities. The structures were grown by molecular beam epitaxy on n-GaSb substrates. The active region core consisted of three GaInAsSb QWs and AlGaAsSb barriers enclosed between 400 nm thick AlGaAsSb broad waveguides. On both sides of the active region 2 µm thick AlGaAsSb cladding layers were grown. For p- and n-type doping we used Be and Te, respectively.

There are two parameters which have to be considered in order to increase the emission wavelength above 2.5 µm: the quantisation energy which should be minimised by choosing wide QWs, and the In-concentration of the GaInAsSb material which should be increased. However, the increase of the In-concentration must be accompanied by an increase of the compressive strain in order to avoid type-II band alignment. This compressive strain in turn limits the maximal QW thickness.

For our 2.7 µm laser we chose a QW thickness of 20 nm to minimize the quantisation energy. To avoid relaxation of the QWs and achieve a good type-I band alignment the As-concentration was chosen to result in a compressive strain of 1.4%. X-ray measurements showed that all other layers were lattice matched to the substrate.

We fabricated ridge waveguide lasers with stripe widths between 3 and 30 µm using standard lithography and reactive ion etching with SiCl₄. For the p- and n-side contacts we used Ti/Pt/Au and Ge/Au/Ni/Au, respectively. Laser bars with different lengths between 500 µm and 1.5 mm were cleaved and mounted on copper heatsinks. The facets of our lasers were uncoated.

Fig.1: Output power against current characteristic for a 1 mm long laser diode with 10 µm stripe width operating CW at 10°C.

Inset: room temperature CW spectrum at 1.2 j_th.

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Figure 1 shows a typical continuous wave (CW) output power against current characteristic. For a laser with 10 µm stripe width and 1 mm length the threshold current is 70 mA at 10°C. The inset of figure 1 depicts the spectrum of this laser at room temperature in CW mode with an emission wavelength of 2.73 µm. The maximum operating temperature in CW mode was 30°C. Pulsed operation was observed up to 60°C, limited only by our experimental setup.

The lasers exhibit low threshold current densities, which can be seen in figure 2: for pulsed operation at 20°C a laser with 1.5 mm cavity length and 30 µm stripe width shows a threshold current density of 356 A/cm². A linear fit of this plot results in an extrapolated threshold current density at infinite length of 79 A/cm² per QW.

The temperature dependence of the threshold current density is shown in figure 3: The lasers were measured in the temperature range between 283 and 313 K. In this temperature range we derived a characteristic temperature of 45 K. Typical GaSb-based lasers in the 2-2.3 µm range show much higher characteristic temperatures but our value is consistent with laser results published for longer wavelengths around 2.6 to 2.7 µm.

In conclusion we fabricated low threshold lasers in the GaInAsSb/AlGaAsSb material system with an emission wavelength of 2.7 µm. The lasers work CW at room temperature and show an extrapolated threshold current density of 79 A/cm² per QW for pulsed operation at room temperature.

Fig.2: Pulsed operation threshold current density per QW against reciprocal cavity length for a laser with 30 µm stripe width (20°C).

Fig.3: Temperature dependence of threshold current density J_th for pulsed operation.
Fabrication of AlGaAsSb/GaInAsSb PIN photodetector with 2.2 µm cut-off wavelength

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Infrared detectors with cutoff wavelength beyond 2µm are very important for trace gas sensing and industrial process control. Because the bandgap of the quaternary alloy GaInAsSb lattice matched to GaSb ranges from 1.7 µm to 4.4 µm, this material is often used as the absorbing layer in photodetectors for the wavelength range from 2 to 3µm operating at room temperature.

The PIN photovoltaic detector structure was grown on n-type GaSb substrate by molecular beam epitaxy. It consists of a 0.25 µm thick, p-type Al0.21Ga0.79As0.02Sb0.98 high bandgap window on top, a 2.5 µm thick not intentionally doped Ga0.81In0.19As0.16Sb0.84 active layer and an n-type 1 µm thick Ga0.81In0.19As0.16Sb0.84 layer. The band diagram of the device structure is shown in fig. 1. The high bandgap AlGaAsSb window was added in order to reduce the surface recombination noise.

The mesa structures with sizes of 250 µm×250 µm, 500 µm×500 µm, and 1 mm×1 mm were formed by photolithography and etched in NaK tartrate:HCl:H2O2:H2O solution. A 250 nm SiO2 layer was sputtered immediately after etching for insulation and junction protection. To planarize the structure, photosensitive polymide (BCB) was spinned on the wafer and patterned by photolithography. For the front p-type contacts, layers of Ti/Pt/Au were evaporated and the n-type contacts were formed by Ge/Au/Ni/Au. Both kinds of contacts were alloyed at 380 °C. After the processing, the devices were separated and mounted in standard TO cans. Fig. 2 shows the scanning electron microscopy photograph of the device structure.

Fig. 3 is the I-V characteristics of a 250 µm×250 µm GaInAsSb detector. The turn-on voltage is around 0.3 V and the reverse current at -0.2 V is only 34 µA. Fig. 4 gives the

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room temperature spectral photoresponse of a 1 mm×1 mm GaInAsSb detector. The cutoff wavelength is 2.24 µm and the maximum responsivity is about 1 A/W which is comparable to commercial available extended InGaAs detectors. The short wavelength cutoff of this detector is 1 µm which is not shown in the figure. By the linear relationship between spectral photoresponse and wavelength

\[ S(\lambda) = \frac{q\eta\lambda}{hc} \]

where \( S(\lambda) \) is the spectral photoresponse, \( q \) is electron charge, \( \eta \) is quantum efficiency, \( \lambda \) is wavelength, \( h \) is Plank constant and \( c \) is the speed of light, a maximum quantum efficiency of 60% can be determined. The Johnson noise limited detectivity can be calculated by

\[ D'_\text{J,L} = \frac{S(\lambda) \cdot (R_0A)^{1/2}}{2 \cdot (kT)^{1/2}} \]

where \( R_0 \) is differential resistance at zero bias, \( A \) the active area, and \( T \) the temperature. For this GaInAsSb photodetector, the Johnson noise limited detectivity is \( 7.7 \times 10^9 \text{ cmHz}^{1/2}/\text{W} \).

In conclusion, high sensitivity GaInAsSb PIN photodetectors with a AlGaAsSb window were fabricated. The cutoff wavelength is 2.24 µm and the responsivity is 1 A/W at room temperature.
InP-based vertical-cavity surface-emitting lasers with emission wavelengths between 1.3µm and 2.0µm

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Vertical-cavity surface-emitting lasers (VCSELs) with long emission wavelengths above 1.3µm are highly desired components for data transmission systems and trace-gas sensing applications. Advantages compared to edge emitting lasers are longitudinal single-mode operation, circular output beam, low beam divergence, and low fabrication cost. In the past, our group has realized an InP-based concept with a buried tunnel junction (BTJ) and electrical pumping (see fig. 1 in “InP-based vertical-cavity surface-emitting lasers for 10Gbit/s optical communication” in this report). Feasible emission wavelengths covered the range from 1.3µm to 1.8µm. Recently, we have been able to extend the upper wavelength limit of our devices to 2.0µm. Devices with emission wavelengths above the telecom wavelength of 1.55µm are mainly important for trace-gas sensing by absorption spectroscopy. To detect a certain species of gas, a laser beam is directed through a chamber which contains the gas to investigate. The wavelength of a VCSEL can be tuned by a few nanometers by current modulation, mainly by heating the small volume device. Thus, transmission is detected vs. wavelength, showing distinct absorption lines depending on the gas contained in the chamber. Absorption lines of interesting species are 1.651µm (CH₄), 1.747µm (HCl), 1.954µm (N₂O), 2.004µm (CO₂), and many absorption lines of H₂O in the vicinities of 1.85µm and 1.98µm. Here, it is crucial to exactly hit the absorption wavelength of the desired gas. On-wafer testing capability and the small wavelength drift across the wafer prove to be considerable advantages of the VCSEL compared to distributed feedback lasers, boosting fabrication yield and reducing cost.

Fig. 1a shows the optical output power vs. the driving current of a device at 2.01µm emission wavelength, along with the normalized emission spectra for different currents illustrating the above-mentioned slight tuning with current in fig. 1b. In this case, emission can be tuned from 2011nm at 2mA to 2014nm at 7mA. The maximum output power reaches 220µW.

Fig. 1: PI-characteristics (a) and normalized emission spectra for several currents of a 2.01µm-VCSEL featuring a tunnel junction diameter of D = 4µm (b).

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Higher output powers are hampered by strong waveguiding in the device. It worsens the suppression of higher transversal modes and impairs the ground-mode output power. Waveguiding in our concept is mainly provided by the higher effective index at the center of the cavity due to the tunnel junction mesa (see fig. 1 in “InP-based vertical-cavity surface-emitting lasers for 10Gbit/s optical communication” in this report). To allow for higher ground-mode output power, the effective index step has to be lowered by reducing the height of this mesa, demanding a thinner $n$-layer in the tunnel junction. On the other hand, if the $n$-layer becomes too thin, the tunneling effect can no longer be retained and the $pn$-junction forms a reverse-biased diode with considerably higher resistance. The limiting thickness depends on the doping concentration which has to be increased to permit even thinner $n$-layers. For these reasons, the influence of the doping levels of both layers in the tunnel junction and the thickness of the $n$-layer on the contact resistivity were investigated. Measurements show that in the investigated regime the resistivity scales almost inversely linear with the doping concentration. Unfortunately, inter-valence-band-absorption in the layers of the tunnel junction rises linearly with the doping concentration, leading to higher internal losses (fig. 2a), and thus forcing a trade-off. If the doping level is fixed at $8 \times 10^{19} \text{cm}^{-3}$, the IV-characteristic stays ohmic down to a certain thickness of the $n$-layer between 6nm and 9nm. Above this critical thickness, contact resistivity changes only slightly with thickness, below no tunneling occurs and a $pn$-diode is formed. This behavior is illustrated in fig. 2b by comparison of the corresponding IV-characteristics.

The effects investigated limit the reduction of waveguiding by decreasing the height of the tunnel junction mesa.

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**Fig. 2:** Contact resistivity vs. doping concentration in the layers of the tunnel junction for thickness of the $n$- and $p$-layer $d_n = 50 \text{nm}$ and $d_p = 20 \text{nm}$, respectively (a), and IV-characteristics of circular tunnel junctions with diameter $D = 12 \mu \text{m}$ and thickness of the $n$-layer $d_n = 6 \text{nm}$ and $9 \text{nm}$, respectively (b).

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**Widely tunable twin-guide lasers**

René Todt, Thomas Jacke, Fabian Köhler, Ralf Meyer, and Markus-Christian Amann

Wavelength-agile single-frequency laser diodes have attracted much attention in recent years. This is mainly due to their large field of application in telecommunications. For present wavelength-division multiplexed (WDM) communication networks, tunable lasers are mainly attractive as backup for conventional distributed feedback (DFB) lasers. However, widely tunable lasers are even more interesting for future generation optical networks. Here they offer the potential to increase the capacity, the functionality and the flexibility by enabling novel optical network architectures.

Although there are already several types of monolithically integrated widely tunable lasers available, practically all of them show various shortcomings. In most designs, characterization and control of the device is quite elaborate. Furthermore, output powers are relatively low. To overcome these issues, a novel widely tunable laser, the so-called widely tunable twin-guide (TTG) laser, is currently being realized at the WSI.

The device consists of three major parts: the active region, the tuning region, and the grating sections. As can be seen from the schematic drawing in Fig. 1a, all three parts are transversally integrated. Hence, a very compact design is possible. Longitudinal single-mode operation is ensured by the use of gratings, which provide wavelength selective distributed feedback. As this filtering mechanism is similar to the one of DFB lasers, an equally high side-mode suppression can be expected. The active and tuning sections consist of pin-junctions that can be biased independently from each other. By applying a forward bias to the tuning diode, carriers are injected into the tuning layer and an electron-hole plasma is generated. Exploiting the free-carrier plasma effect, the effective refractive index of the structure is decreased, which in turn results in a shift of the Bragg wavelength of the gratings and thereby the emission wavelength is tuned.

It is important to note, that there are two grating sections, that can be tuned independently and that provide comb-like reflection spectra with differing peak spacings (Fig. 1b). Lasing will occur where reflection peaks of both gratings overlap. Hence, rather large

![Fig. 1: a) Longitudinal cross-section of a widely tunable twin-guide laser employing sampled gratings. b) Simulations of the corresponding grating reflection spectra.](image)

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Wavelength changes can be achieved by keeping the reflection spectrum of one grating section fixed while tuning the other one (Vernier-effect tuning).

The grating design is one of the decisive points with respect to device performance. Sampled gratings (SGs) as well as superstructure gratings (SSGs) provide a comb-like reflection spectrum (Fig. 2) and can be employed. Sampled gratings consist of a periodic arrangement of corrugated and uncorrugated areas. The corrugated area usually makes up about 10% of one sampling period. Hence, the grating reflectivity is fairly low. Furthermore, the reflectivity is dropping off rapidly away from the Bragg-wavelength, which is highly undesirable. However, SGs can be conveniently fabricated by a combination of holography and optical lithography. On the other hand, SSGs consist of an arrangement of several superperiods. Within each superperiod the grating period is varied continuously or in discrete steps. Since there is no uncorrugated area, the reflectivity is high. Moreover, the reflection spectrum is almost flat within the relevant wavelength range. Therefore, SSGs provide a better performance than SGs. However, time-consuming electron-beam lithography is necessary for the definition of SSGs.

On the technological side, various key fabrication steps have been investigated. As example, Fig. 3 shows a GaInAsP grating that has been embedded into InP. Fabrication included electron-beam lithography for the definition of the grating into PMMA resist, reactive ion etching to transfer the pattern into the GaInAsP layer and overgrowth with InP by chemical beam epitaxy.

Since the widely tunable twin-guide laser is to be realized as a buried-heterostructure device, also the corresponding fabrication technology was extensively investigated. Details thereto can be found in the article “Buried Heterostructure Lasers based on Chemical Beam Epitaxy” in this issue.

**Fig. 2:** Schematic structure and simulations of the reflection spectra of a) sampled gratings and b) superstructure gratings.

**Fig. 3:** SEM image of the stained cross-section of a GaInAsP grating (period 240 nm) embedded into InP.
Micro-electro-mechanically tunable two-chip VCSELs for 1.55 µm

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Widely tunable long-wavelength vertical-cavity surface-emitting lasers (VCSELs) are highly attractive light sources for telecom applications within Dense-Wavelength-Division-Multiplexing (DWDM) systems. Fixed wavelength laser replacement, single transverse-mode operation with high side mode suppression, and high coupling efficiency into an optical fiber are some of the main benefits of these devices. Various monolithic concepts of micro-electro-mechanically tunable VCSELs have been presented to date. Here we investigate a new approach to electrically pumped, tunable VCSELs based on a two-chip concept: The device (Fig. 1) consists of a “half-VCSEL” and a movable top mirror membrane. The active part of the device is very similar to the design of the long-wavelength VCSEL that has been developed at the WSI during the last few years except that the front mirror has been omitted (see report by R. Shau). In a first epitaxial run AlInAs is deposited on a 2-inch InP substrate followed by the active region and a Buried Tunnel Junction (BTJ) consisting of heavily n- and p-doped GaInAs layers. The active region is formed by 5 compressively strained AlGaInAs quantum wells of 8nm thickness. The diameter of the BTJ and therefore the current confinement aperture is then defined by dry chemical etching (ECR-RIE) just before the InP overgrowth (CBE). Finally a dielectric bottom mirror consisting of 2.5 pairs of CaF₂/a-Si and an Al₂O₃-antireflection coating (AR) is deposited after substrate removal.

The micro-mechanically tunable mirror membrane is an MBE-grown (In₀.₀₃)GaAs/AlGaAs-Bragg-mirror of about 19.5 periods leading to a theoretical reflectivity of 99.7%. After defining the shape of the membrane and partial removal of the semi-

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insulating GaAs substrate the additional 3% Indium contents in some of the GaAs-layers leads to a stress-gradient and therefore to a rotation-symmetric concave curvature which is essential for a stable resonator. A picture of such a membrane mirror is shown in Fig. 2.

![Fig. 2: Top view of a mirror membrane chip with four flexible suspension beams.](image)

![Fig. 3: Profile measurement of a membrane chip showing the curvature of the mirror due to the stress gradient caused by In-incorporation.](image)

The membranes are characterized by white-light interference spectroscopy (Fig. 3) showing radii of curvature in the range from 3 to 20mm depending on the size of the mirror. Deflection of the membrane, i.e. wavelength tuning, can be achieved by electrothermal heating of the thin suspension beams by injecting a small current using the via-hole contacts through the substrate (Fig. 1 and 2). First below-threshold tuning experiments with this kind of device confirm the possibility of wide tuning over about one free spectral range (FSR) as can be seen in Fig. 4. This is an essential feature for a 2-chip concept since one has to compensate for arbitrary misalignments after assembling the two parts.

![Fig. 4: Below-threshold emission spectra for different tuning currents injected into the membrane.](image)

One major difference of this device compared to the well-established non-tunable VCSEL arises from the fact that the top mirror is not planar but curved. This leads to certain stability requirements in terms of a very good match of resonator length, membrane curvature, and BTJ-diameter. Especially the latter one can be estimated to be 20µm assuming a gaussian beam. By a careful adjustment of these parameters single mode laser devices should be possible to fabricate since a 20µm BTJ-VCSEL has already been demonstrated.

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InP-based vertical-cavity surface-emitting lasers for 10 Gbit/s optical communication

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Surface-emitting laser diodes (VCSELs) are light sources with attractive features such as low power consumption, relaxed fiber coupling and small footprint. Today, fiberoptical communication at Gbit/s rates per channel is divided into \( \lambda \approx 0.85 \mu m \) systems for short reach utilizing GaAs-based VCSELs and long-wavelength systems around 1.31 \( \mu m \) or 1.55 \( \mu m \) for metro access and long haul networks with InP-based edge-emitting laser diodes. To comply with telecom market requirements, long-wavelength VCSELs should be capable of direct modulation up to 10 Gbit/s preferably at 1.55 \( \mu m \). This could be demonstrated for the first time with VCSELs employing the concept of a buried tunnel junction (BTJ) in conjunction with a hybrid Bragg reflector. **Structure:** The device as shown in Fig. 1 consists of an \( n \)-doped InGaAlAs/InAlAs distributed Bragg reflector which is lattice-matched to InP, an active region with five compressively strained InGaAlAs quantum wells and a \( p^{++}\)-InGaAlAs/\( n^{++}\)-InGaAs tunnel interface, all grown by solid-source MBE. The tunnel region is patterned with an ECR-RIE etch process and regrown by gas-source CBE with \( n\)-InP, which serves as a heat and current spreader, improving the overall electrical resistance, which is a crucial parameter for the \( RC \) modulation bandwidth. A short-period dielectric layer stack (CaF\(_2\)/a-Si) in conjunction with a gold termination yielding a reflectivity of 99.75 % is placed in contact to a plated gold pad that diverts heat and provides mechanical stability.

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Results: Single-mode devices with a reduced contact pad capacitance have been mounted on SMA connectors. A bias-T was used to mix a constant forward bias current with pulses of a $2^7-1$ word length pseudo-random bit sequence pattern. Reasonably opened eyes have been obtained (Fig. 3) and bit error rates down to $10^{-12}$ have been measured up to 10 Gbit/s (Fig. 4), where measurements were limited by detector responsivity. As can be seen in Fig. 2, the device remains transversally single-mode under modulation, so the system bandwidth is limited by chromatic dispersion of the broadened fundamental mode in standard single-mode fibers. As a figure of merit, the bandwidth-distance product may be estimated to 389 GHz·km which allows for 10 Gbit/s over 40 km in a metro area network. The high-speed capabilities may be further improved by replacing the SiO$_2$ insulating layer by a polymer coating. We gratefully acknowledge the support of F. Mederer with the measurements.

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Quantum dot cascades

Nicolaus Ulbrich, Jochen Bauer, Ruhan Boy, Giuseppe Scarpa, Dieter Schuh, and Gerhard Abstreiter

Quantum cascade lasers are reliable coherent light sources in the mid-infrared spectral region enabling important advances in areas such as ultrahigh resolution spectroscopy, ultrahigh sensitivity gas-sensing and optical communications. Despite significant developments such as terahertz lasing and high-temperature pulsed operation, quantum cascade lasers exhibit high threshold current densities due to the short non-radiative intersubband lifetimes. A significant reduction of the threshold current density is expected in quantum dot cascade lasers exploiting the properties of a zero-dimensional density of states. Another possible advantage over conventional quantum cascade lasers is the possibility to fabricate vertical-cavity surface emitting lasers with its intrinsic properties such as low power consumption and the prospect of low-cost fabrication.

We have developed quantum dot cascade structures in the material system GaAs/AlGaAs on GaAs substrates which are based on optical transitions from AlInAs quantum dots. The detailed layer design is displayed in Figure 1 where one active section and two injection superlattices are displayed at an applied bias of 50 kV/cm. The superlattice is based on GaAs quantum wells and Al_{0.45}Ga_{0.55}As barriers with a conduction band offset of 340 meV. The active region consists of two 4.8 and 5.4 nm thick quantum wells and one layer of self-assembled Al_{0.5}In_{0.5}As quantum dots grown on top of a 1.1 nm thin barrier and capped by a 6.9 nm thick tunneling barrier. The upper sublevel s of the optical transition is formed by the s-shells of the quantum dot layer.

The self-assembled quantum dots are obtained using the Stranski-Krastanov growth mode during the deposition of 11.5 monolayers of highly strained Al_{0.5}In_{0.5}As on Al_{0.45}Ga_{0.55}As at a growth rate of 0.2 Å/s and a growth temperature of 490°C. The atomic force microscopy topology of a sample containing a stack of 11 quantum dot layers is displayed in the inset of Figure 2. Each quantum dot layer is embedded in a 45.4 nm thick Al_{0.45}Ga_{0.55}As matrix to investigate the conditions as obtained in quantum dot cascade structures. We have obtained lens shaped quantum dots with high dot densities of ~1000 dots/µm^2 and base diameters of ~20 nm. The average height is ~3 nm. The photoluminescence measurements at 4.2 K show a broad peak centered at ~1.82 eV and an inhomogeneous linewidth of ~83 meV which is due to an increase of the quantum dot lateral and...
vertical size when growing several quantum dot layers. The latter is due to an incremental build up of strain induced by the highly strained lower quantum dot layers. For the electroluminescence measurements we have grown 10 active periods of the bandstructure displayed in Figure 1 at a growth temperature of 490°C on top of an n-type semi-insulating GaAs substrate, an 800 nm thick n'-GaAs contact layer and a 200 nm thick n' GaAs buffer layer grown at 630°C. The growth is terminated with a 500 nm n'-GaAs buffer layer and a 200 nm n'-GaAs contact layer. The growth of each active period starts with the quantum well superlattice followed by the deposition of 11.5 monolayer Al0.5In0.5As which are directly capped with the 6.9 nm Al0.45Ga0.55As barrier to prevent In desorption. After the growth sequence the samples are processed by wet chemical etching into 120×120 µm² mesa with Ti/Pt/Au top and bottom metallization. The samples are then lapped under an angle of 45° for coupling out the light. The electroluminescence spectra of two adjacent mesas obtained with a Fourier transform infrared spectrometer using step-scan and lock-in detection technique at a temperature of 80 K are displayed in Figure 3. The spectra reveal mid-infrared intraband electroluminescence in the range of 150 meV. We believe that the spectra are due to transitions between the s-shells of the quantum dot layers and the subband 2 of the GaAs quantum wells. The broadening at higher currents is due to the inhomogeneous broadening of the quantum dot ensemble.

In conclusion we have developed suitable quantum dots for cascaded intraband emission and demonstrated mid-infrared intraband electroluminescence from a cascade of coupled AllnAs quantum dots and GaAs quantum wells. The reported design may be a possible solution to obtain stimulated emission from quantum dot cascade structures.

Fig. 2: Photoluminescence spectrum at 4.2K of a stack of 10 Al0.5In0.5As quantum dot layers embedded in 45.4 nm Al0.45Ga0.55As. Inset: atomic force microscopy topology of the topmost uncapped quantum dot layer.

Fig. 3: Low-temperature electroluminescence spectra from a cascade containing a stack of 10 Al0.5In0.5As quantum dot layers embedded in an Al0.45Ga0.55As/GaAs superlattice.

supported by: Bayerische Forschungsstiftung
the electroluminescence measurements have been performed in collaboration with Stefan Schmult and Werner Wegscheider from University of Regensburg
ASV frequency tripler for mm-wave applications

Markus Krach¹, Jürgen Freyer, and Manfred Claassen

The applications for mm-wave and THz devices and the resulting power sources are in rapid expansion, not only in the field of radio astronomy, atmospheric sensing and radar systems but particularly also in the range of medical electronics and surveillance systems.

In order to achieve reasonable output power at frequencies above 200 GHz, frequency multipliers or even multiplier chains based on varactor diodes are used, which are pumped by fundamental oscillators. To realize high-performance frequency multipliers, two main criteria have to be fulfilled simultaneously: optimization of the varactor structure as well as matching of active device and circuit at both, input and output frequency. The anti-serial Schottky varactor (ASV) used as non-linear device for frequency tripling consists of two inhomogeneously doped GaAlAs/GaAs Schottky diodes in anti-serial connection which are quasi-monolithically integrated into a microstrip circuit on quartz substrate. Because of its symmetric structure there is no need for DC bias circuitry and idler circuits at even harmonics. Therefore, the rf-circuit can be substantially simplified as compared to non-symmetric varactor devices. The performance, i.e. cut-off frequency and conversion efficiency of the tripler device are mainly determined by the elastance change and the series resistance of the varactor. Corresponding design and optimization criteria for the single Schottky diodes are deduced from a theoretical model of the ASV which describes both, the static and the dynamic behavior of the device and which includes relevant physical effects such as self-biasing and maximum drift velocity of electrons in the modulation zone of the varactor. A large elastance change can be realized with optimized inhomogeneously doped Schottky diodes, the layer sequence of which is given in Fig. 1 a). In order to minimize the series resistance of the ASV, a device structure has been developed which enables a combination of large area n-contacts and small active areas (see Fig. 1 b)). The former leads to a negligible contact resistance and therewith to a reduced total series resistance of the ASV. Small area devices are preferred for high frequency operation, since for given circuit losses the conversion efficiency increases with decreasing diode area and additionally, impedance matching can be achieved more easily.

**Fig. 1:** Layer sequence and schematic device structure of a ASV

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The circuit of the microstrip frequency tripler including the ASV is realized quasi-monolithically integrated on quartz substrate. The initial material for the diodes is grown on GaAs substrate by MBE technique and is bonded upside-down on a quartz substrate with a thermocompression technique. After the removal of the GaAs substrate, the entire rf-circuit including the varactor is fabricated on quartz substrate with standard photoresist technology. Due to the high reproducibility of this technique, an accurate control of the whole circuit is possible taking into account also the parasitic elements of the device mounting. The knowledge of circuit impedances as well as varactor impedances at both, input and output frequency, is absolutely necessary since impedance mismatch is the main source for the degradation of frequency multipliers. The total circuit consisting of the nonlinear ASV, a low-pass filter, input as well as output coupling (see Fig. 2) is simulated and optimized by the help of commercial software such as HFSS and ADS. The impedance of the ASV is calculated from the elastance-voltage and current-voltage characteristic of the device. The circuit impedance can be adjusted and optimized by varying widths and lengths of the individual microstripline components.

The tripler-circuit on the quartz chip is tested in a split waveguide mount with two symmetrical halves (see Fig. 2). Fine-tuning of the tripler is obtained by backshorts at both, input and output waveguide. Experimental results of the ASV frequency tripler confirm the theoretically predicted excellent performance showing a flange to flange conversion efficiency of 22 % and an output power of 15 mW at 228 GHz for an ASV with a device area of \( A = 6 \cdot 10^{-7} \text{ cm}^2 \). This represents an essential improvement in efficiency compared to other tripler concepts which are normally based on hetero-barrier varactors (HBV) and which show maximum efficiencies up to 13 % (Meola et al., TUM) or 11 % (Mélique et al., USTL). The ASV shows a significantly higher power capability than typical HBVs. This enables smaller device diameters at given input power resulting in higher efficiency and makes the ASV a promising candidate for mm-wave and sub-mm wave applications.

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supported by DFG (SFB 348 (B14))
Charge storage in InGaAs/GaAs quantum dots: wavelength selectivity and absorption mechanisms

M. Kroutvar, Y. Ducommum, J. J. Finley, M. Bichler, and G. Abstreiter

In this project, we are working on a memory device based on charge storage in self-assembled InGaAs/GaAs quantum dots (QDs). Our approach to create a device for memory application is based on a recent proposal by Muto [1] in which the inhomogeneously broadened absorption spectrum of a QD ensemble is used for wavelength domain information storage. Resonant optical excitation below the GaAs band edge only generates charge in a subset of the QD ensemble for which the transition energies of the QDs match the excitation energy. This enables us to probe absorption mechanisms in self-assembled QDs.

Our charge storage device demonstrates for the first time direct optical readout of the stored charge distribution in the spectral domain. This device consists of a single layer of MBE grown InGaAs QDs embedded within the intrinsic region of either an n- or p-type Schottky photodiode for hole or electron storage respectively. The operating principle of the hole storage device (analogous for electron storage) is divided into two parts. In the write mode, the device is reverse biased $V_{\text{write}}$. Following optical excitation of the QD sub-ensemble ($\hbar\omega_{\text{write}}$), photo-generated excitons are rapidly ionized by the strong vertical electric field. Whilst electrons readily tunnel out of the dot, the holes remain stored by virtue of an AlGaAs barrier adjacent to the QD layer. Optical readout of the stored charge distribution is achieved by application of a forward bias pulse $V_{\text{read}}$, resulting in non-resonant injection of majority charge carriers into the QD layer and subsequent neutralization of the stored charge by radiative recombination. The resulting charge storage spectrum is then expected to directly reflect the spectral distribution of stored charge within the QD ensemble.

Resonant optical excitation pulses were delivered using a mechanically modulated (~20 kHz) continuous-wave Ti-sapphire laser ($\hbar\omega_{\text{write}} = 1.25-1.37$ eV) or a Ti-sapphire in combination with a laser diode ($\hbar\omega_{\text{write}} = 1.27$ eV) for two-color experiments. The charge storage time ($\Delta t$) is defined as the time delay between the switching off of the optical write and the onset of the electrical readout pulses.

Fig. 1a,b shows typical charge storage spectra for electron (a) and hole (b) devices for $\Delta t = 12$ µs. For comparison, the non-resonantly excited PL spectrum ($\Gamma_i \sim 80$ meV) of the electron storage device is also plotted (dashed line in Fig. 1a). The overall form of the charge storage spectra were found to be very similar for both samples, consisting of two principle components: a sharp peak, labeled $R$, very close to the excitation energy and a much broader band, labeled $S$, ~ 20-50 meV to lower energy. The characteristic shape of the charge storage spectra presented in Fig. 1a,b indicates the existence of two distinct absorption channels. We attribute peak $R$ to direct, selective charging of QD ground states addressable by the excitation laser. For such ground states, energy relaxation cannot occur and consequently $R$ arises at an energy ($E_R$) close to $\hbar\omega_{\text{write}}$ (see Fig. 1d). The broader, low energy emission band $S$ is attributed to dots that are quasi-resonantly excited via their excited state. During this process, photo-generated excitons relax into their ground state before ionization occurs, resulting in a luminescence band shifted below $\hbar\omega_{\text{write}}$ by the energy separation between ground and excited states ($E_{\text{QD,exc}} - E_{\text{QD}}$, see Fig. 1e). In addition, phonon-assisted excitation (Fig. 1f) may also contribute to $S$, due to enhanced exciton-LO phonon coupling in QDs as evidenced by the barely resolved satellite features shown in Fig. 1a,b (see arrows), shifted by $\Delta E_{\text{LO}} \sim 32$ meV below $\hbar\omega_{\text{write}}$.

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To investigate the role of these different processes, electron storage spectra were recorded as $\hbar \omega_{\text{write}}$ was tuned from 1.30 to 1.24 eV, corresponding to shifting the excitation from the peak of the excited state absorption to energies for which only ground state absorption exists (Fig. 1c). As $\hbar \omega_{\text{write}}$ shifts to lower energy the amplitude of the emission sideband (S) reduces strongly as $\hbar \omega_{\text{write}}$ is shifted below the excited state absorption edge at ~ 1.25 eV. As a consequence, for all the spectra where $\hbar \omega_{\text{write}} < 1.25$ eV in Fig. 1c, peak R is the dominant emission feature. These observations are consistent with the dominant contribution to S - arising from quasi-resonant charge storage via excited states.

The investigation of the carrier dynamics during the charge storage shows no temporal development of the storage spectrum within our time window of 25 µs indicating that the storage lifetime is much longer than 25 µs at $T = 10$ K.

Finally, in order to demonstrate wavelength selective charge storage we performed two-color experiments, applying a sequence of optical pulses during each excitation cycle. The first pulse, delivered by a diode laser, was centered at $\hbar \omega_{\text{write}} \sim 1.27$ eV whilst the second pulse from the Ti:sapphire, at $\hbar \omega_{\text{write}}$ was tuned over the ground state absorption spectrum. Charge storage spectra after $\Delta t = 12$ µs are presented in Fig. 2. For both excitation energies the emission spectrum has the form described above, consisting of a resonant peak R close to each excitation energy and a weaker low energy side band S. These results demonstrate unambiguously that two separate and distinct sub-ensembles are charged by the two excitation frequencies, reaffirming the conceptual basis for implementation of wavelength division multiplexing for information storage using inhomogeneously broadened media such as QDs.

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supported by DFG (SFB 348 B1)
2D-1D-2D resonant tunneling transistor

Sebastian Roth, Frank Ertl\textsuperscript{1}, Dieter Schuh, Max Bichler, and Gerhard Abstreiter

Among nano-structured semiconductors, resonant tunneling devices attract much attention due to their negative differential resistance (NDR). Applied research uses this quantum mechanical property to the generation of high frequency waves. Basic research obtains valuable information about quantum tunneling and transport, which is contained in detail in the device characteristics. We are concerned with the fundamental tunneling processes between low-dimensional systems, for which we present latest results.

We developed the resonant tunneling transistor shown in Fig. 1, in which we study lateral tunneling between large 2D electron reservoirs and one 1D quantum wire. The basic GaAs/AlGaAs heterostructure that hosts the 2D system is a 750 nm long vertical transistor realized by means of the cleaved-edge overgrowth technique. In addition the transistor is equipped with two 3 nm wide AlAs tunneling barriers with a distance of 12 nm, which serve to confine the 1D state. By application of a gate voltage the 1D quantum wire and the adjoining 2D systems can be electrostatically induced and the carrier density tuned.

We examine this device by measuring the $I-V$ output characteristic under gate voltage bias at 4.2 $\text{K}$ with typical results introduced in Fig. 2. Although the curves assemble partially of gate leakage currents and pure transistor behavior, clear signatures related to resonant tunneling are found. The typical NDR region ($B$) and its replica ($\tilde{B}$) at negative voltages are identified with tunneling from the 2D source through the 1D ground state. The weaker resistance change ($C$) is attributed to tunneling through the first excited 1D state. We also observe a shift of these features to larger absolute source-drain voltages with decreasing gate voltage. This can be explained qualitatively by treating the device as a series of two transistors and a tunnel junction. In particular the sole tunneling region is visible only in the limit of large gate voltages and small source-drain bias. Concerning that part of the parameter space we then conclude that the position of the resonances marginally varies with gate voltage. We understand this behavior well from a 2D-1D tunneling model including energy and momentum conservation. A NDR feature then occurs when the bottom of an 1D state drops below the conduction band edge of the 2D source. This condition is also independent of Fermi energy (or gate voltage, respectively).

The experimental results and schematics of the important tunneling conditions are included.

\textbf{Fig. 1:} Schematic cross section of this novel 2D-1D-2D resonant tunneling transistor.

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summarized in Fig. 3. We calculate the conductance from the $I$-$V$ curves to highlight the tunneling resonances in the color coded plot for the whole source-drain/gate voltage space. The black lines mark the resonances (B) and (C) mentioned before, and moreover (A) appears, designating resonant filling of the 1D ground state from the 2D source. This picture includes all replica lines $(\overline{A}), (\overline{B})$ and $(\overline{C})$ at negative bias, which indicates the high symmetry in the 2D-1D-2D resonant tunneling transistor structure.

**Fig. 2:**
$I$-$V$ traces of this device for several gate voltages. Resonant tunneling features appear onto the curve (B) and (C), which possess replica ($\overline{B}$) at negative bias. The region near (B) or ($\overline{B}$) exhibits a clear NDR.

**Fig. 3:** Color coded conductance plot of the source-drain/gate voltage parameter space. The black lines are guides to the eyes of the characteristic tunneling signatures. Line (A) corresponds to the beginning of resonant filling of the 1st 1D subband from the 2D source. Line (B) to the end of this case followed by the NDR region and line (C) shows resonant filling by the 2nd 1D subband. $(\overline{A}), (\overline{B})$ and $(\overline{C})$ are negative bias replica lines.

supported by BMBF (01BM912) and DFG (SFB 348)
Efficient computational method for ballistic current and application to quantum logic gates
M. Sabathil, D. Mamaluy, and P. Vogl

We have developed a highly efficient method to calculate the ballistic transmission and current through an arbitrarily shaped, multi-terminal two- or three-dimensional open device. The method is applicable to all cases where the current is sufficiently small so that a ballistic model is meaningful and charge self-consistency is not relevant, such as quantum dot devices, quantum wires, or interferometer type of structures.

So far, the computational effort for the calculation of the ballistic current through a realistic 3D device of complex geometry has been out of reach since standard methods such as the quantum transmitting boundary method [QTBM] scale with the third power of number of nodes or grid points of the entire device and thus require the repeated inversion of matrices of typical size ~ $10^6 – 10^8$.

We have shown that the calculation of the energy dependent transmission function can be reduced to a single calculation of a small percentage of the stationary states in the isolated device plus the inversion of a small matrix that is energy dependent. The size of this matrix is shown to be governed by the size of the boundary region between the leads and the device. This robust method that we term contact block reduction method (CBR) drastically reduces the effort for the calculation of the ballistic transmission from an $N^3$ to an $N^2$ problem. This enables us to investigate the transmission through large 2D and 3D devices including realistic strain and potential profiles for the first time.

Recently we extended the CBR method to multi-band $k.p$-Hamiltonians which allows us to investigate structures with highly non-parabolic bands as they occur in quantum wells for example. The use of an atomic tight-binding basis would be also of high interest for calculations regarding molecular structures and transport in carbon nanotubes for example.

One useful application of the CBR method is the field of quantum information. Bertoni et al. proposed a scheme that uses two quantum wires as a Qbit and windows between the wires as simple logic gates. We calculated the ballistic current through a Mach-Zender type interferometer based on an AlGaAs/GaAs high mobility 2DEG with top gates to induce lateral confinement (Fig. 1). This device should be suitable for experimental realization since the dimensions of the lateral structure are large enough

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on the order of 50 nm and the electrostatic confinement via the gates offers the necessary flexibility to compensate for errors in the fabrication.

The potential profile of the 3D structure is calculated in equilibrium with applied gate voltages using our device simulator nextnano\textsuperscript{3}. Then the CBR method is used to calculate the ballistic transmission through the device which leads to the current via the well known Landauer formula. In this way we show that the coherent current through the two wire system can be controlled and manipulated via simple electrostatic gates with voltages in the order of few mV.

Figure 2 shows the calculated current between the leads L1-L3 and L1-L4. These two channels correspond to either no action on the Qubit (L1-L3) or an inversion of the single Qubit (L1-L4). The parameters used for this calculation are within the range of experiments, like a temperature of 300mK, a Fermi level of 8 meV and an applied Bias voltage 0.1 mV.

The results show strong oscillations of the current between the two channels which is due to the phase shift induced by the gate G2. Thus even small gate voltages below 1mV are enough to switch the quantum logic gate.

This example emphasises the usefulness of the CBR algorithm in terms of designing quantum logic devices based on ballistic transport of electrons or holes, which should enable experimentalists to built these structures far more precisely than without previous simulations.

\textbf{Fig. 2:} This graph shows the current between leads L1-3 vs. L1-4 as a function of applied gate voltage at the gate G2 (Fig. 1) which acts as phase shifter for the upper channel. For a gate voltage between 0 and 2mV, the current oscillates four times between the two channels, which corresponds to a phase shift of $\pi$ for 0.5 mV gate voltage.
Fabrication of AlInAs/InGaAs heterojunction bipolar transistors using self-aligned process

Rajendra Singh¹, Markus Krach, Jürgen Freyer, and Gerhard Böhm

Heterojunction bipolar transistors (HBTs) based upon III-V semiconductors are potentially very useful in high-speed and microwave integrated circuits. In an HBT, the emitter is composed of a wide-bandgap semiconductor while the base is made up of low-bandgap material. For npn transistors this lead to band offsets at the heterointerface that favours the injection of electrons into the base while retarding the hole injection into the emitter. This allows the base to be more heavily doped than the emitter resulting in low base resistance and emitter-base capacitance, both of which are required for high frequency operation.

The earlier works on HBTs focused more on the AlGaAs/GaAs based material system. But it is proved beyond doubt that InP-based HBTs have numerous advantages. For the AlInAs/InGaAs material to be investigated the low bandgap InGaAs is favorable since it shows higher values of electron mobility, peak electron velocity and saturated electron velocity as compared to GaAs leading to lower parasitic resistances and shorter transit times. Additionally, InGaAs, which is used for the base, has a surface recombination velocity which is approximately three orders of magnitude lower than for GaAs. This leads to significantly lower 1/f noise behaviour, favourable for the use in low-phase-noise oscillator circuits.

In the present investigation the \( \text{Al}_{0.48}\text{In}_{0.52}\text{As}/\text{In}_{0.53}\text{Ga}_{0.47}\text{As} \) material system has been used. The bandgap discontinuity between \( \text{Al}_{0.48}\text{In}_{0.52}\text{As} \) (\( E_g = 1.48 \) eV) and \( \text{In}_{0.53}\text{Ga}_{0.47}\text{As} \) (\( E_g = 0.76 \) eV) is about 0.72 eV, having a conduction-band offset \( \Delta E_C \) of 0.52 eV and valence-band offset \( \Delta E_V \) of 0.20 eV. The layer sequence for the transistor structure with abrupt emitter-base junction is shown in Fig. 1. The individual layers are grown lattice-matched by MBE technique on semi-insulating InP (100)-substrates. Since for an abrupt junction of \( \text{Al}_{0.48}\text{In}_{0.52}\text{As} \) and \( \text{In}_{0.53}\text{Ga}_{0.47}\text{As} \) there exists a spike in the conduction band which affects the electron flow, epitaxial material with graded junction from emitter to base has also been fabricated. For this purpose a superlattice structure of InAlAs/InGaAs having total thickness of 48 nm and doped n-type up to \( 5 \times 10^{17} \) cm\(^{-3} \) is inserted between the emitter and base layers. The superlattice structure is comprised of three parts, each part having 8-period superlattice with thickness of 2.0 nm per period. The thickness of InAlAs and InGaAs layers in each period of the three parts are (starting from base side): first part, 0.5 nm and 1.5 nm, second part, 1.0 nm and 1.0 nm, and third part, 1.5 nm and 0.5 nm, respectively.

**Fig. 1: Epitaxial layer structure for the abrupt emitter-base junction HBT**

<table>
<thead>
<tr>
<th>Layer Name</th>
<th>Material Type</th>
<th>Thickness (nm)</th>
<th>Doping ((\text{cm}^{-3}))</th>
</tr>
</thead>
<tbody>
<tr>
<td>Emitter cap</td>
<td>InGaAs n(^{++})</td>
<td>75</td>
<td>( 5 \times 10^{19} )</td>
</tr>
<tr>
<td>Emitter cap</td>
<td>AllInAs n(^{++})</td>
<td>45</td>
<td>( 5 \times 10^{19} )</td>
</tr>
<tr>
<td>Emitter</td>
<td>AllInAs n(^+)</td>
<td>100</td>
<td>( 5 \times 10^{17} )</td>
</tr>
<tr>
<td>Base</td>
<td>InGaAs p(^+)</td>
<td>80</td>
<td>( 5 \times 10^{19} )</td>
</tr>
<tr>
<td>Collector</td>
<td>InGaAs n</td>
<td>700</td>
<td>( 1 \times 10^{16} )</td>
</tr>
<tr>
<td>Subcollector</td>
<td>InGaAs n(^{++})</td>
<td>1000</td>
<td>( 5 \times 10^{19} )</td>
</tr>
<tr>
<td>Substrate</td>
<td>InP SI</td>
<td>--</td>
<td></td>
</tr>
</tbody>
</table>

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Test structures for the HBTs were fabricated using a self-aligned triple mesa-etch process. This technique minimizes the distance between base and emitter contacts and moreover, isolation of the individual devices can also be achieved. The etching of the various epitaxial layers in order to reach down to base, subcollector and SI-InP substrate was performed using selective and nonselective wet chemical etch solutions. E-beam evaporated Ti/Au (20 nm/200 nm) is used to make ohmic contacts to the emitter cap, base and subcollector. The individual metal contacts serve as etch masks for the wet chemical etching of the corresponding mesa structures.

DC current-voltage measurements were performed on the devices by the help of an HP 4142B modular DC source/monitor and Agilent’s IC-CAP as data acquisition and analysis software. The Gummel plot and output characteristics for an abrupt junction HBT are displayed (blue curves) in Fig. 2a and Fig. 2b, respectively. The current gain $\beta$ of the transistor in the high current region is about 15. However, there is an intersection of the emitter and base current which limits the application for base-emitter voltages above 1 V. This intersection is possibly due to the outdiffusion of Be from base to emitter layer and/or due to the defects at the emitter-base interface. The offset voltage and knee voltage, as extracted from the output characteristics, are 0.65 V and 0.90 V, respectively.

![Gummel plots and common-emitter $I_C$-$V_{CE}$ characteristics for abrupt (blue curves) and graded (red curves) junction AlInAs/InGaAs HBTs having emitter dimensions of 50x50 $\mu$m$^2$. In case of (b) the base current $I_B$ is varied from 0 to 500 $\mu$A in steps of 100 $\mu$A for both the HBTs.](image)

Fig. 2: (a) Gummel plots and (b) common-emitter $I_C$-$V_{CE}$ characteristics for abrupt (blue curves) and graded (red curves) junction AlInAs/InGaAs HBTs having emitter dimensions of 50x50 $\mu$m$^2$. In case of (b) the base current $I_B$ is varied from 0 to 500 $\mu$A in steps of 100 $\mu$A for both the HBTs.

Gummel plot and output characteristics for the graded junction transistor are also shown in Fig. 2 (red curves). The current gain for this structure is about 30 for high currents, i.e. roughly twice as compared to the abrupt junction transistor and there is no intersection between base and emitter current in the Gummel plot. Similarly, the offset voltage and knee voltage are also reduced and values of 0.15 V and 0.40 V, respectively, are achieved. The improvements in the transistor performance for the graded junction HBTs due to the smoothening of conduction band at emitter-base junction are evident. Now the efforts for fabrication of small dimension HBTs (emitter: 2x10 $\mu$m$^2$ to 5x20 $\mu$m$^2$) are underway.
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.

**Funded Projects**

1. **Bundesministerium für Bildung und Forschung (BMBF)**
   - Kompetenzzentrum Nanoclub Lateral: „Selbstorganisiertes Wachstum auf Si“ (SOWASI, partner: DaimlerChrysler Ulm)
   - Kompetenzzentrum Nanop: „Anwendung von Nanostrukturen in der Optoelektronik“
   - Laserkristallisation und Strukturierung von mikrokristallinen Halbleitern für photovoltaische Anwendungen
   - Förderschwerpunkt Elektronenkorrelation und Dissipationsprozesse in Halbleiterquantenstrukturen: „Vertikale Ultrakurzkanal- und Tunneltransistoren“, „Deterministische Ladungsquelle“
   - Förderschwerpunkt MultiTeraNet; “Mikromechanisch abstimmbare VCSEL für den Wellenlängenbereich 1,5 µm
   - Förderschwerpunkt neue Materialien; “Materialien für Infrarotkonverter“

2. **Deutsche Forschungsgemeinschaft (DFG)**
   - Sonderforschungsbereich „Nanometer-Halbleiterbauelemente“ (SFB 348)
   - Sonderforschungsbereich „Bioorganische Funktionssysteme auf Festkörpern“ (SFB 563)
   - Schwerpunktprogramm „Gruppe III-Nitride und ihre Heterostrukturen“
   - Schwerpunktprogramm „Silizium-Chemie“
   - Schwerpunktprogramm „Quanten-Hall-Systeme“
   - Schwerpunktprogramm „Optische Übermittlungsverfahren in der Informationstechnik“
   - Forschungskooperation „FORMEDIAN“
   - Am 101/6-1: „Transversal integrierte abstimmbare Laserdioden“
   - Br 1960/1-3: „Lichtemission von Intrabandübergängen in Si/SiGe-Heterostrukturen“
   - Ne 524/1-2: „Mikromechanische und elektronische Diamantanwendungen: CVD-Diamant“
   - Stu 139/6-3: „Amorphe Suboxide des Siliziums“
   - Vo 483/4-3: „Herstellung und Untersuchung epitaktischer Metall-Isolator Resonant-Tunneling-Bauelemente für die Silizium-Nanoelektronik“

3. **Bayerische Forschungsstiftung**
   - Laseroptische Sensorik für die Gasanalytik
   - Langfristprogramm Neue Werkstoffe
4. **European Union**
   - COLLECT (Collective electronic states in nanostructures: INFM and other partners, Italy)
   - DODD's (Doped Diamond Devices and Sensors, partners: Kings College London, Instituut voor Maerial Onderszoek Limburg, Institute of Physics - Acad. of Sciences of the Czech Republic Praha, LEPES, Grenoble, University College London, CEA/LETI Gif sur Yvette, Technion Haifa)
   - ESRQC (Electron Spin Resonance Quantum Computing, partner: INFM-MDM, Italy)
   - NEWTON (New Widely Tunable Lasers for Optical Networks, partners: IMEC, INTUNE, ADC)

5. **Office of Naval Research**
   - N00014-01-1-0242: “Modeling and simulation of semiconductor nanostructures”

6. **Istituto Nazionale per la Fisica della Materia, Italy**
   - Electrically detected magnetic resonance (partner: Laboratorio Materialie Disposiviti per la Microelettronica, Milano)

7. **Industry**
   - Infineon, München, Germany: „Zweidimensionale quantenmechanische Simulation“, „Quantenmechanische MOSFET-Simulation“
   - EADS Deutschland GmbH, Ottobrunn, Germany: „Sensorik auf der Basis von Halbleitern mit großer Bandlücke“
   - Osram Optosemiconductors, Regensburg, Germany: „Prozessierung von GaN Leuchtdioden“
   - Fujitsu Laboratories of Europe LTD., Hayes Park, UK: “Bio-Nanotechnology”

**Other collaborations involve:**

RWTH Aachen, Arizona State University, Universität Augsburg, MBI Berlin, TU Berlin, Universität Bremen, CALTECH, Cambridge University, Wacker Chemitronic Burghausen, University Chicago, Cornell University, TU Darmstadt, Intune Dublin, Universität Duisburg, Universität Erlangen-Nürnberg, FHG IPM Freiburg, Vertilas Garching, WMI Garching, Gayton Photonics (UK), IMEC Gent, Universität Gießen, Universität Göttingen, Universität Graz, TU Hamburg-Harburg, Universität Hamburg, Universität Hannover, FZ Jüllich, Universität Kaiserslautern, Universität GH Kassel, Universität Köln, Linköping University, ETH Lausanne, Universität Leoben, Universität Linz, University College London, Université Montpellier 2, LMU München, Infineon München, Rhode & Schwarz München, Siemens München, Bell Labs. Murray Hill, University of Nagoya, Laser Components Olching, University Osaka, EADS Deutschland Ottobrunn, Universität
Paderborn, Czech Academy of Sciences Prag, Universität Regensburg, Università Rom II Tor Vergata, UCSB Santa Barbara, CL SENES Sofia, ADC Stockholm, MPI FKF Stuttgart, Universität Stuttgart, University of Tokyo, National Institute of Materials Science Tsukuba, NEC Tsukuba, DaimlerChrysler Ulm, Universität Ulm, Naval Research Laboratory Washington, TU Wien, Kistler Instrumente Winterthur, Nanoplus Würzburg, Universität Würzburg, ETH Zürich, Nortel Zürich.
6. Publications

Enhancement of photoluminescence from near-surface quantum dots by suppression of surface state density

Vertical-cavity surface-emitting laser diodes for telecommunication wavelength
M.-C. Amann, M. Ortsiefer, R. Shau, and J. Rosskopf

Surface-emitting laser diodes for telecommunications
M.-C. Amann, M. Ortsiefer, and R. Shau

Long-wavelength InP-based VCSELs
M.-C. Amann

Pyroelectric properties of Al(In)GaN/GaN hetero- and quantum well structures

Optical properties of single charge tuneable InGaAs quantum dots

Level bleaching in a single quantum dot observed by photocurrent spectroscopy
E. Beham, A. Zrenner, F. Findeis, M. Bichler, and G. Abstreiter
Physica E 13, 139-142 (2002)

Nonlinear optical response of highly energetic excitons in GaAs: Microscopic electrodynamics at semiconductor interfaces
M. Betz, G. Göger, A. Leitenstorfer, M. Bichler, G. Abstreiter, and W. Wegscheider

Femtosecond optical response of exciton – LO phonon quasiparticles in GaAs
M. Betz, G. Göger, A. Leitenstorfer, R. Zimmermann, M. Bichler, W. Wegscheider, and G. Abstreiter
phys. stat. sol. (b) 231, No.1, 181-186 (2002)
Two-color femtosecond spectroscopy of blue-shifted InAs/AlGaAs quantum dots
M. Betz, S. Trumm, A. Leitenstorfer, E. Beham, H. Krenner, M. Bichler, A. Zrenner, and G. Abstreiter

InP-based VCSEL technology covering the wavelength range from 1.3 to 2.0 µm
G. Böhm, R. Shau, R. Meyer, M.-C. Amann, M. Ortsiefer, J. Rosskopf, and F. Mederer

Midinfrared intersubband electroluminescence of Si/SiGe quantum cascade structures

Novel Si/Ge quantum dot mid-infrared photodetector structures with in-plane transport
D. Bougeard, K. Brunner, and G. Abstreiter

Anomalous dispersion of charged excitons in dilute two-dimensional electron systems at low temperatures
K. B. Broocks, P. Schroter, D. Heitmann, C. Heyn, C. Schuller, M. Bichler, and W. Wegscheider

Ge quantum dots in Si: self-assembly, stacking and level spectroscopy

Si/Ge nanostructures
K. Brunner

Epitaxial growth of phosphorus doped diamond on {111} substrate
N. Casanova, A. Tajani, E. Gheeraert, E. Bustarret, J. A. Garrido, C. E. Nebel, and M. Stutzmann
Diamond and Related Materials 11, 328-331 (2002)

A quantum dot infrared photodetector with lateral carrier transport
L. Chu, A. Zrenner, D. Bougeard, M. Bichler, and G. Abstreiter
Miniband transport in vertical superlattice field effect transistors
R. A. Deutschmann, W. Wegscheider, M. Rother, M. Bichler, and G. Abstreiter

Electronic properties of antidot lattices fabricated by atomic force lithography
A. Dorn, M. Sigrist, A. Fuhrer, T. Ihn, T. Heinzel, K. Ensslin, W. Wegscheider, and M. Bichler

Magnetotransport through AFM-defined antidot arrays
A. Dorn, M. Sigrist, A. Fuhrer, T. Ihn, T. Heinzel, K. Ensslin, W. Wegscheider, and M. Bichler

Properties of grain boundaries in laser-crystallized silicon thin films
C. Eisele, T. Bach, C. E. Nebel, and M. Stutzmann

Ge-Si nanostructures for quantum-effect electronic devices

Vertical field effect transistors realized by cleaved-edge overgrowth
Physica E 13, 920-924 (2002)

Device characteristics of vertical field effect transistors with ultra-short InGaAs/GaAs channels
F. Ertl, R. A. Deutschmann, D. Schuh, M. Bichler, and G. Abstreiter

Fine structure of charged and neutral excitons in InAs-Al_{0.6}Ga_{0.4}As quantum dots
J. J. Finley, D. J. Mowbray, M. S. Skolnick, A. D. Ashmore, C. Baker, A. F. G. Monte, and M. Hopkinson

Lateral variation of the electron density across a Hall bar in high electric and magnetic fields
J. Frankenberger, A. Zrenner, M. Bichler, and G. Abstreiter
Intraband absorption and photocurrent spectroscopy of self-assembled p-type Si/SiGe quantum dots

Intersubband transitions of boron-doped self-assembled Ge quantum dots
T. Fromherz, W. Mac, C. Miesner, K. Brunner, G. Bauer, and G. Abstreiter

Electronic properties of nanostructures defined in Ga(Al)As heterostructures by local oxidation
A. Fuhrer, A. Dorn, S. S. Luscher, T. Heinzel, K. Ensslin, W. Wegscheider, and M. Bichler

Energy spectra of quantum rings
A. Fuhrer, S. Luscher, T. Ihn, T. Heinzel, K. Ensslin, W. Wegscheider, and M. Bichler
Microelectronic Engineering 63, 47-52 (2002)

Spin-sensitive bleaching and monopolar spin orientation in quantum wells
S. D. Ganichev, S. N. Danilov, V. V. Bel’kov, E. L. Ivchenko, M. Bichler, W. Wegscheider, D. Weiss, and W. Prettl

Circular photogalvanic effect in SiGe semiconductor quantum wells

Removal of spin degeneracy in p-SiGe quantum wells proved by spin photocurrents

Capacitance-voltage studies of Al-Schottky contacts on hydrogen-terminated diamond
J. A. Garrido, C. E. Nebel, and M. Stutzmann

Electrical and optical measurements of CVD diamond doped with sulfur
J. A. Garrido, C. E. Nebel, M. Stutzmann, E. Gheeraert, N. Casanova, and E. Bustarret

Fabrication of in-plane gate transistors on hydrogenated diamond surfaces
J. A. Garrido, C. E. Nebel, R. Todt, G. Rösel, M.-C. Amann, M. Stutzmann, E. Snidero, and P. Bergonzo
Characterization of sub-micron in-plane devices in H-terminated diamond
J. A. Garrido, C. E. Nebel, M. Stutzmann, G. Rösel, R. Todt, M.-C. Amann, E. Snidero, and P. Bergonzo
phys. stat. sol. (a) 193, 517-522 (2002)

n-Type doping of diamond by sulfur and phosphorus
E. Gheeraert, N. Casanova, A. Tajani, A. Deneuville, E. Bustarret, J. A. Garrido, C. E. Nebel, and M. Stutzmann

Electron spin resonance of phosphorus in n-type diamond
T. Graf, M. S. Brandt, C. E. Nebel, M. Stutzmann, and S. Koizumi
phys. stat. sol.(a) 193, 434-441 (2002)

The Mn^{3+/2+} acceptor level in group III nitrides
T. Graf, M. Gjukic, M. S. Brandt, M. Stutzmann, and O. Ambacher

Low threshold 2.72µm GaInAsSb/AlGaAsSb multiple-quantum-well laser
M. Grau, C. Lin, and M.-C. Amann

Spectral measurement of the hall angle response in normal state cuprate superconductors
M. Grayson, L. B. Rigal, D. C. Schmadel, H. D. Drew, and P.-J. Kung

The lever-arm model: describing resonant tunneling under bias at a fractional quantum Hall edge
M. Grayson, D. C. Tsui, L. N. Pfeiffer, K. W. West, and A. M. Chang
Physica E 12, 80-83 (2002)

Nonequilibrium band structure of nano-devices
S. Hackenbuchner, M. Sabathil, J. A. Majewski, G. Zandler, P. Vogl, E. Beham, A. Zrenner, and P. Lugli

Structural and optical properties of vertically correlated Ge island layers grown at low temperatures

Magnetotransport in freely suspended two-dimensional electron systems for integrated nanomechanical resonators
E. M. Höhberger, R. H. Blick, F. W. Beil, W. Wegscheider, M. Bichler, and J. P. Kotthaus
Tunneling in the quantum Hall regime between orthogonal quantum wells
M. Huber, M. Grayson, M. Rother, R. A. Deutschmann, W. Biberacher, W. Wegscheider, M. Bichler, and G. Abstreiter
Physica E 12, 125-128 (2002)

Local spectroscopy of edge channels in the quantum Hall regime with local probe techniques
T. Ihn, J. Rychen, T. Vancura, K. Ensslin, W. Wegscheider, and M. Bichler
Physica E 13, 671-674 (2002)

Scanning gate measurements on a quantum wire
T. Ihn, J. Rychen, T. Cilento, R. Held, K. Ensslin, W. Wegscheider, and M. Bichler
Physica E 12, 691-694 (2002)

Tunneling between edge channels in the quantum Hall regime manipulated with a scanning force microscope
T. Ihn, J. Rychen, K. Ensslin, W. Wegscheider, and M. Bichler

Er3+ luminescence in a-SiOx:H
A. Janotta, M. Schmidt, R. Janssen, Ch. Buchal, and M. Stutzmann

Dependence of the doping efficiency on material composition in n-type a-SiOx:H

Role of defect centers in recombination processes in GaN monocrystals
N. V. Joshi, A. Cros, A. Cantarero, H. Medina, O. Ambacher, and M. Stutzmann

Femtosecond intersubband scattering of holes in Si1-xGex/Si quantum wells

Aharonov-Bohm oscillations of a tuneable quantum ring
U. F. Keyser, S. Borck, R. J. Haug, M. Bichler, G. Abstreiter, and W. Wegscheider

Integrating suspended quantum dot circuits for applications in nanomechanics
J. Kirschbaum, E. M. Hohberger, R. H. Blick, W. Wegscheider, and M. Bichler
Intrinsic amorphous and microcrystalline silicon by hot-wire-deposition for thin film solar cell applications
S. Klein, F. Finger, R. Carius, H. Wagner, and M. Stutzmann

Microcrystalline silicon by hot-wire CVD for thin film solar cell applications
S. Klein, J. Wolff, F. Finger, R. Carius, H. Wagner, and M. Stutzmann

Intrinsic microcrystalline silicon by hot-wire chemical vapour deposition for solar cell applications
S. Klein, F. Finger, R. Carius, O. Kluth, L. Baia Neto, H. Wagner, and M. Stutzmann
2nd HWCVD, Denver (2002)

High efficiency thin film solar cells with intrinsic microcrystalline silicon prepared by hot wire CVD
S. Klein, F. Finger, R. Carius, B. Rech, L. Houben, M. Luysberg, and M. Stutzmann

Effect of the interface on the local structure of Ge-Si nanostructures

Local structure of Ge quantum dots self-assembled on Si(100) probed by x-ray absorption fine-structure spectroscopy
A. V. Kolobov, H. Oyanagi, S. Wei, K. Brunner, and G. Abstreiter

Improved frequency tripler design with anti-serial Schottky varactor
M. Krach, J. Freyer, and M. Claassen

Finite wavevector scattering on the $v = \frac{2}{3}$ huge longitudinal resistance
S. Kraus, J. G. S. Lok, W. Dietsche, K. v. Klitzing, W. Wegscheider, and M. Bichler
Physica E 12, 72-75 (2002)

From Quantum Hall ferromagnetism to huge longitudinal resistance at the 2/3 fractional Quantum Hall state

Spin polarized tunneling through single-crystal GaAs (001) barriers
S. Kreuzer, J. Moser, W. Wegscheider, D. Weiss, M. Bichler, and D. Schuh
**Femtosecond buildup of Coulomb screening in a photoexcited electron-hole plasma**  
A. Leitenstorfer, R. Huber, F. Tauser, A. Brodschelm, M. Bichler, and G. Abstreiter  

**Negative magneto-drag of double layer 2DEGs**  

**Morphological transformation of In$_y$Ga$_{1-y}$As islands, fabricated by Stranski-Krastanov growth**  
A. Lorke, R. Blossey, J. M. Garcia, M. Bichler, and G. Abstreiter  

**Quantum wires and quantum dots defined by lithography with an atomic force microscope**  
S. Luscher, A. Fuhrer, R. Held, T. Heinzl, K. Ensslin, M. Bichler, and W. Wegscheider  

**Heterostructure field effect transistors based on nitride interfaces**  
J. A. Majewski, G. Zandler, and P. Vogl  

**Efficient light emission at 1.54 µm from Er in Si excited by hot electron injection through thin suboxide layers**  
M. Markmann, A. Sticht, F. Bobe, G. Zandler, K. Brunner, G. Abstreiter, and E. Müller  

**Thermoresistive and piezoresistive properties of Wurtzite N-GaN**  
S. Mingiacchi, P. Lugli, A. Bonfiglio, G. Conte, M. Eickhoff, O. Ambacher, A. Rizzi, A. Passaseo, P. Visconti, and R. Cingolani  
phys.stat.sol. (a) **190**, 281-286 (2002)

**Reliability of Pt and Ni based Schottky contacts on Al$_{0.31}$Ga$_{0.69}$N**  

**Resonant effect of Zener tunneling current**  
M. Morifuji, T. Imai, C. Hamaguchi, A. Di Carlo, P. Vogl, G. Böhm, G. Tränkle, and G. Weimann,  

**Low temperature properties of the p-type surface conductivity of diamond**  
Observation of ion-induced changes in the channel current of high electron mobility AlGaN/GaN transistors (HEMT)
R. Neuberger, G. Müller, M. Eickhoff, O. Ambacher, and M. Stutzmann

Hole transport in SiGe channels on step-bunched vicinal Si surfaces
R. Neumann, K. Brunner, and G. Abstreiter

High-speed data transmission with 5.5µm vertical-cavity surface-emitting laser
M. Ortsiefer, R. Shau, F. Mederer, R. Michalzik, J. Rosskopf, G. Böhm, F. Köhler, C. Lauer, M. Maute, and M.-C. Amann
Post-deadline Session 4, 28th European Conference on Optical Communication, (12.09.2002), Per Danielsen, Com. Technical University of Denmark

High-speed modulation up to 10 Gbit/s with 1.55 µm wavelength InGaAlAs VCSELs
M. Ortsiefer, R. Shau, F. Mederer, R. Michalzik, J. Rosskopf, G. Böhm, F. Köhler, C. Lauer, M. Maute, and M.-C. Amann

Long-wavelength InP-based vertical-cavity surface-emitting lasers
M. Ortsiefer and M.-C. Amann

Long-wavelength InP-based VCSELs
M. Ortsiefer, R. Shau, and M.-C. Amann

Manipulation of the homogeneous linewidth of an individual In(Ga)As quantum dot

Nanotechnology for SAW devices on AlN epilayers
T. Palacios, F. Calle, E. Monroy, J. Grajal, M. Eickhoff, O. Ambacher, and C. Prieto

Ultrafast switch-off of an electrically pumped quantum-dot laser

L-valley electron transport in GaAs-AlAs double-barrier resonant tunneling structures studied by ballistic electron emission microscopy
D. Rakoczy, G. Strasser, C. Strahberger, and J. Smoliner
Local oxidation of hydrogenated diamond surfaces for device fabrication
B. Rezek, J. A. Garrido, M. Stutzmann, C. E. Nebel, E. Snider, and P. Bergonzo
phys. stat. sol. (a) 193, 523-528 (2002)

Laser beam induced currents in polycrystalline silicon thin films prepared by interference laser crystallization
B. Rezek, C. E. Nebel, and M. Stutzmann

Towards fully quantum mechanical 3D device simulation
M. Sabathil, S. Hackenbuchner, J. A. Majewski, G. Zandler, and P. Vogl
J. of Computational Electronics 1, 81 (2002)

Full-band approaches for the quantum treatment of nanometer-scale MOS structures
F. Sacconi, M. Povolotskyi, A. Di Carlo, P. Lugli, M. Städele, C. G. Strahberger, and P. Vogl

High-performance 5.5 µm quantum cascade lasers with high-reflection coating
G. Scarpa, N. Ulbrich, J. Roßkopf, G. Böhmer, G. Abstreiter, and M.-C. Amann

Improved large optical cavity design for 10.6 µm (Al)GaAs quantum cascade lasers
G. Scarpa, N. Ulbrich, A. Sigl, M. Bichler, D. Schuh, M.-C. Amann, and G. Abstreiter

Strain-compensated InP-based quantum cascade lasers with and without injector regions
G. Scarpa, N. Ulbrich, G. Böhmer, G. Abstreiter, and M.-C. Amann

Low-loss GaInAs-based waveguides for high-performance 5.5 µm InP-based quantum cascade lasers
G. Scarpa, N. Ulbrich, G. Böhmer, G. Abstreiter, and M.-C. Amann

Gas sensitive GaN/AlGaN-heterostructures
J. Schalwig, G. Müller, M. Eickhoff, O. Ambacher, and M. Stutzmann

Group III-nitride-based sensors for combustion monitoring
J. Schalwig, G. Müller, O. Ambacher, and M. Stutzmann
Hydrogen response mechanism of Pt-GaN Schottky diodes

Nonlinear superlattice transport limited by Joule heating
R. Scheuerer, K. F. Renk, E. Schomburg, W. Wegscheider, and M. Bichler

Anomalous x-ray diffraction on InAs/GaAs quantum dot systems
T. Schulli, M. Sztucki, V. Chamard, T. Metzger, and D. Schuh

Anomalous x-ray diffraction on InAs/GaAs quantum dot systems
T. Schulli, M. Sztucki, V. Chamard, T. Metzger, and D. Schuh

Band-gap renormalization of modulation-doped quantum wires
S. Sedlmaier, M. Stopa, G. Schedelbeck, W. Wegscheider, and G. Abstreiter

InP-based vertical-cavity surface-emitting lasers with high output power and large modulation bandwidth
R. Shau, G. Böhm, F. Köhler, M.-C. Amann, M. Ortsiefer, and J. Rosskopf

Gate-voltage control of spin interactions between electrons and nuclei in a semiconductor

Silicon/silicon suboxide heterostructures grown by molecular beam epitaxy
A. Sticht, M. Markmann, K. Brunner, G. Abstreiter, and E. Müller

Thin SiOx layers embedded in single crystalline silicon
A. Sticht, M. Markmann, K. Brunner, and G. Abstreiter

GaN-based heterostructures for sensor applications

Inter- and intra-subband LO phonon emission rates in GaAs/AlGaAs quantum disks
N. Suzumura, M. Yamaguchi, N. Sawaki, and P. Vogl
Probing the phonon dispersion relations of graphite from the double-resonance process of Stokes and anti-Stokes Raman scatterings in multiwalled carbon nanotubes

Two-dimensional electron-gas actuation and transduction for GaAs nanoelectromechanical systems
H. X. Tang, X. M. H. Huang, M. L. Roukes, M. Bichler, and W. Wegscheider

A novel split gate design to study interaction effects in quantum wires
M. Tornow, M. Heiblum, D. Mahalu, H. Shtrikman, and V. Umansky
Physica E 13, 89-93 (2002)

Acoustical and optical magnetoplasma excitations in a bilayer electron system

Intersubband staircase laser
N. Ulbrich, G. Scarpa, G. Böhm, G. Abstreiter, and M. C. Amann

Multiscale approaches for metal thin film growth
P. Vogl, U. Hansen, and V. Fiorentini

Magnetoresistance calculations for a two-dimensional electron gas with unilateral short-period strong modulation
K. Výborný, L. Smrcka, and R. A. Deutschmann

Ultrafast intersubband scattering of holes in p-type modulation-doped Si₁₋ₓGeₓ/Si multiple quantum wells

Coherent properties of a two-level system based on a quantum-dot photodiode
A. Zrenner, E. Beham, S. Stufler, F. Findeis, M. Bichler, and G. Abstreiter

Optically detected single-electron charging in a quantum dot
A. Zrenner, F. Findeis, M. Baier, M. Bichler, G. Abstreiter, U. Hohenester, and E. Molinari
Physica E 13, 95-100 (2002)
7. **Invited Talks**

**Gerhard Abstreiter**

1. *Silizium basierende Quantenbauelemente und Selbstorganisation*  
   GMM Workshop, Integrierte Silizium-Hetero-Bauelemente, Infineon München, Germany (25.04.2002)

2. *Quanteninformationsverarbeitung: Herausforderungen für die Nanotechnologie*  
   NanoDE Kongress, Bonn, Germany (07.05.2002)

3. *Control of charge and spin in novel semiconductor nanostructure devices*  
   Kolloquium, PDI Berlin, Germany (31.05.2002)

4. *Control of charge and spin in novel semiconductor nanostructure devices*  
   10th Int. Symposium, Nanostructures: Physics and Technology, St. Petersburg, Russia (17.06.2002)

5. *Control of charge and spin in semiconductor nanodevices*  
   NATO Advanced Research Workshop on Frontiers of Spintronics and Optics in Semiconductors, Boston, USA (21.06.2002)

6. *SiGe-from science to application*  
   Kolloquium, Universität Linz, Austria (12.10.2002)

7. *Kontrolle von Ladung und Spin in Nano-Bauelementen: Basis für Quanteninformationstechnologie in Halbleitern*  
   Kolloquium, LMU München, Germany (21.10.2002)

8. *Kontrolle von Ladungen und Spins in Halbleiter-Nanostrukturen*  
   Kolloquium, Universität Bremen, Germany (28.11.2002)

**Markus-Christian Amann**

1. *Surface-emitting laser diodes for telecommunications*  
   Second Joint Symposium on Opto- and Microelectronic Devices and Circuits  
   Stuttgart, Germany (10.-16.03.2002)

2. *Vertical-cavity surface-emitting laser diodes for telecommunication wavelength*  
   Semiconductor Lasers and Optical Amplifiers for Lightwave Communications Systems (SPIE), Boston, USA (29.-30.07.2002)

3. *Long-wavelength InP-based VCSELs*  
   5th International Conference on Mid-Infrared Optoelectronic Materials and Devices  
   Annapolis, Maryland, USA (08.-11.09.2002)
4. *Wavelength-tunable laser diodes for optical communications*

Evelin Beham

1. *Single quantum dot photodiodes: Coherent excitation of a two-level system with electric contacts*
   12th International Winterschool on New Developments in Solid State Physics
   Low-dimensional Systems: From 2D to molecules, Mauterndorf, Austria (28.02.02)

Martin S. Brandt

1. *Spins in Halbleitern: Zustände mit Zukunft*
   Seminar, Universität Paderborn, Germany (19.02.2002)

2. *Spins in Silizium: Zustände mit Zukunft*
   Seminar, Universität Cottbus, Germany (11.04.2002)

3. *Spins: Spione in Materie*
   Seminar, Universität Konstanz, Germany (30.04.2002)

4. *Flipping spins in small places*
   Oberseminar, Center for Nanoscience, Ludwig-Maximilians-Universität München,
   Germany (5.07.2002)

5. *DX-centers in AlN and AlGaN alloys*
   10th Int. Conference on Shallow Level Centers in Semiconductors (SLCS),
   Warschau, Poland (25.07.2002)

Karl Brunner

1. *Selbstorganisation und Quanteneffekte in Si/Ge Nanostrukturen*
   Seminar, Universität Linz, Austria (21.01.2002)

2. *Quanteneffekte in Si/Ge Nanostrukturen*
   GMM-Workshop Infineon München, Germany (25.04.2002)

3. *Selbstorganisation und Quanteneffekte von Si/Ge Nanostrukturen*
   Seminar, Universität Würzburg, Germany (16.05.2002)

4. *IR-Sensoren und Solarzellen mit Si/Ge Nanostrukturen*
   Workshop „Analyse der Nanotechnologie-Anwendungen in Raumfahrtentwicklungen und –systemen“ der DLR Köln, Germany (04.06.2002)
Martin Eickhoff

1. *Pyroelectric GaN and AlGaN for surface sensitive sensors*
   Interdisziplinäres Seminar, Lehrstuhl für allgemeine Materialwissenschaft, Universität Kiel (Germany) (29.04.2002)

2. *Novel sensor applications of III-nitrides*
   XIIth Semiconducting and Insulating Materials Conference (SIMC-XII), Smolenice Castle, Slovakia (02.07.2002)

3. *Sensor applications of pyroelectric GaN and AlGaN*
   Swedish Sensor Center, Universität Linköping, Sweden (06.10.2002)

4. *Field effect sensors based on group III-nitrides*
   Physics Department University of Orlando, USA (09.12.2002)

Jonathan Finley

1. *Spectroscopic investigations of individual quantum dots*
   Seminar, TU-München, Physik-Department E16, Germany (27.06.2002)

2. *Perturbation spectroscopy of zero dimensional semiconductor nanostructures*
   Seminar, TU-München, Physik-Department E11, Germany (12.11.2002)

3. *Optically probing quantum states in individual quantum dots*
   Seminar, Universität Würzburg, Germany (27.11.2002)

Matthew Grayson

1. *Momentum conserved tunneling into integer quantum Hall edges*
   Seminar, LMU München, Germany (29.04.2002)

2. *Tunneling into a sharp quantum Hall edge*
   Seminar, Scuola Normale Superiore Pisa, Italy (02.05.2002)

3. *Spectral measurement of the Hall angle in normal state cuprate superconductors*
   Seminar, WMI Garching, Germany (11.06.2002)

4. *T- and L-junction quantum wells: New device structures for investigating quantum Hall edges*
   Nanophase Conference, Erice, Italy (20.07.2002)

5. *Spectral measurement of the Hall angle in normal state cuprate superconductors*
   Seminar, MPI Stuttgart, Germany (28.10.2002)
6. **T- and L-junction quantum wells: New device structures for investigating quantum Hall edges**
   Seminar, Universität Karlsruhe, Germany (11.11.2002)

**Jacek A. Majewski**

1. **First-principles studies of Rashba effect in semiconductor superlattices**
   XIV Ural International Winter School on the Physics of Semiconductors,
   Ekaterinburg, Russia (18-22.02.2002)
2. **Heterostructure field effect transistors based on nitride interfaces**
   Workshop of European Centre of Excellence ‘What’s new in nitrides?’, Warsaw,
   Poland (05-07.07.2002)

**Giuseppe Scarpa**

1. **Low-loss GaInAs-based waveguides for high-performance 5.5 µm InP-based quantum cascade lasers**
   5th International Conference on Mid-Infrared Optoelectronic Materials and Devices
   Annapolis, Maryland, USA (08.-11.09.2002)

**Robert Shau**

1. **InP-based high-speed vertical-cavity surface-emitting lasers at 1.55 µm**
   SPIE Photonics Fabrication Europe 2002: VCSELs and Optical Interconnects,
   Brügge, Belgium (31.10.2002)

**Martin Stutzmann**

1. **Gruppe III-Nitride: Von der Optoelektronik zur Sensorik**
   Kolloquium Universität Erlangen-Nürnberg, Germany (22.01.02)

2. **Thin film silicon for bioelectronics**
   1st ASINET Workshop, Salerno, Italy (07.03.02)

3. **Sensitive electrons in wide bandgap semiconductors**
   CMD19, Brighton, UK (10.04.2002)

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