Development of a III-nitride electro-optical modulator for UV–vis

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We present first experimental results of a novel electro-optical modulator for UV–vis based on III-nitrides. This device consists of a modulation layer which changes its complex dielectric constant influenced by the electric field and a field-dependent carrier distribution. Here, we fabricated and investigated different configurations for the modulation layer structure. The core of the structures was a p–i–n structure in which the i-layer was either GaN or AlGaN. Furthermore, a conventional undoped AlGaN/GaN heterostructure containing a two-dimensional electron gas was grown. Samples were characterized by high-resolution X-ray diffraction (2theta-omega scans and reciprocal space mappings), atomic force microscopy and capacitance–voltage measurements; the reflectivity modulation was determined using a UV–vis white-light reflectometer and a DC voltage source. The results show that reflectivity modulation is readily achieved, caused by the control of the charge carrier density and the electric field. We obtained a relative change in reflectivity of ±1.5% from −8 V to +2 V. © 2019 The Japan Society of Applied Physics

1. Introduction

Electro-optical modulators (EOM) have found widespread application in optical data transmission and spectroscopy. For the infrared spectrum, well-performing solutions which can be integrated into integrated circuits are available.1 Yet, in the UV–vis (ultraviolet–visible) range, only external discrete EOM can be used. Most of them, like Pockels cells, are slow and require hundreds of volts for operation. Here, we present a novel concept of an EOM for UV and visible light, which is based on nitride semiconductors.

Various optical effects have been analyzed in the III-nitrides, like the temperature dependence of the refractive indices of III-nitrides,2–4 the quantum-confined Stark effect in quantum structures,5–8 the Franz–Keldysh effect (FKE) in GaN p–n junctions or Schottky barriers,5,10–16 and the opposing effects of Burstein–Moss shift (BMS) and bandgap renormalization (BGR) in GaN with different free electron densities.7,15,17–22 Depending on the device structure and doping profile, several of these effects can affect the reflectivity simultaneously when a voltage is applied.

If such an electrically controllable layer is integrated into a resonant structure like a distributed Bragg reflector (DBR), the stopband of the device can be shifted. This will furthermore amplify the changes in reflectivity (or transmittivity) when optical properties of this particular layer are modulated.23 In this work, we focused on fabrication and characterization of such a controllable layer. We investigated three different structures: a GaN p–i–n structure of which the electric field and the width of the space charge region can be modulated, an undoped AlGaN/GaN heterostructure containing a polarization-induced 2DEG,13,14,24–26 and an undoped AlGaN/GaN heterostructure containing a two-dimensional electron gas (2DEG/2DHG), and an undoped AlGaN/GaN heterostructure containing a two-dimensional electron gas (2DEG).

2. Experimental details

All samples were grown by MOVPE in an AIXTRON 200/4 RF-S reactor with in situ reflectometry and pyrometry on 2 inch (0001) sapphire with 0.3° offcut towards m-plane, beginning with an AlN/GaN buffer. This buffer layer starts with an approximately 30 nm thick AlN nucleation at 770 °C with subsequent annealing and growth of 270 nm AlN at 1270 °C, followed by a 2 μm thick GaN buffer at 1060 °C. Afterwards one of the three active structures were grown: sample GaN-PIN, a GaN p–i–n structure as reference in which 600 nm nGaN with an electron concentration of \( n = 7 \times 10^{18} \text{cm}^{-3} \), 40 nm uidGaN and 200 nm pGaN with a free hole concentration of \( p = 7 \times 10^{17} \text{cm}^{-3} \) were successively grown, sample AlGaN-PIN a p–i–n structure in which the i-GaN was replaced with uid-AlGaN, and sample 2DEG-HS a 2DEG containing undoped GaN/AlGaN heterostructure with a sheet carrier concentration of \( n_s = 1 \times 10^{13} \text{cm}^{-2} \) (see Fig. 1). All charge carrier concentrations were determined on reference samples via Hall measurements. The activation of Mg acceptors in samples GaN-PIN and AlGaN-PIN was performed at 700 °C for 900 s in nitrogen by rapid thermal annealing (RTA) to prevent Mg diffusion into the i-layer.27–28 To fabricate test devices, metals were deposited by e-beam evaporation to form electrical contacts. For all samples, after a recess by reactive ion etching down to the bottom layer, Ti/Al/Ni/Au stacks were deposited for n-contacts and were alloyed by RTA for 30 s at 825 °C under nitrogen atmosphere. Afterwards, semi-transparent contacts (8 nm Ni/8 nm Au) were deposited on top of the structures. For samples GaN-PIN and AlGaN-PIN, the semi-transparent p-contacts were alloyed for 600 s at 525 °C under ambient nitrogen atmosphere. X-ray diffraction to determine thickness and composition (2theta-omega scan) and to verify pseudomorphic growth (reciprocal space mapping). Also, atomic force microscopy (AFM) was performed to analyze surface morphology. Devices were electrically characterized by capacitance–voltage (CV) and current–voltage (IV) measurements before the modulation of white-light reflectance was measured by an
UV–vis reflectometer equipped with a DC voltage source. Additionally, a heater was included into the reflectometer setup to perform temperature-dependent measurements. The optical properties of the semi-transparent contacts (deposited on blank sapphire) were determined separately with a UV–vis spectrometer.

3. Results and discussion

Reciprocal space mappings of samples AlGaN-PIN and 2DEG-HS [shown in Figs. 2(b) and 2(c), respectively] verify pseudomorphic growth. By fitting simulations to 2theta-omega scans of samples AlGaN-PIN and 2DEG-HS [see Fig. 2(a)], a barrier thickness of 43 nm and a composition of Al0.24Ga0.76N was determined for sample AlGaN-PIN and a barrier thickness of 21 nm and a composition of Al0.25Ga0.75N for sample 2DEG-HS.

CV measurements shown in Fig. 3 reveal a p–i–n diode-like profile for samples GaN-PIN and AlGaN-PIN. This indicates that the i-region is depleted in the investigated voltage range and the depletion region expands with negative bias voltages. Sample 2DEG-HS shows the step-like characteristic curve of a 2DEG which is completely depleted at −6 V. This characteristic curve is absent in sample AlGaN-PIN, indicating that any polarization induced 2DEG or 2DHG at the lower or upper AlGaN interfaces are fully depleted in the observed voltage range. In this voltage range, the current does not exceed 10 mA for all samples, and the maximum electrical loss is 17 mW (Fig. 3). This will cause a negligible amount of Ohmic heating, which will be discussed later.

Figure 4 shows the reflectivity spectra of all three samples. The oscillations and the envelopes in these spectra originate from Fabry–Pérot interferences in the buffer structure. The reflectivity of samples GaN-PIN and AlGaN-PIN is multiple times lower than that of sample 2DEG-HS. The cause is the difference in surface morphology of the top GaN layers which was found by AFM. The root mean square (RMS) of the surface roughness of sample 2DEG-HS equals 1.9 nm, whereas the surfaces of samples GaN-PIN and AlGaN-PIN are much rougher with a RMS roughness of 3.9 nm because
of the negative effect of high Mg concentrations on the surface morphology. The transparency of the semi-transparent contacts was estimated to 22% for wavelengths between 250 to 550 nm by measuring a reference contact sample on blank sapphire.

When a bias voltage is applied to the sample for electro-optical modulation, leakage current can raise the temperature of the device and influence the optical properties and therefore the modulation spectrum. To estimate the effect of Ohmic heating and to distinguish this from isothermal electro-optical modulation, temperature-dependent measurements of reflectivity were performed. Figure 5 shows the relative reflectivity change due to temperature changes; a rise of the temperature from 25 °C to 30 °C changes the reflectivity by up to 2%, while the changes in sample 2DEG-HS remain below 2% even at 50 °C. For sample 2DEG-HS, the cause of thermo-optical modulation are most likely only the shift of Fabry–Pérot resonances by thermal expansion and the temperature dependence of the complex refractive indices of the nitrides. For sample AlGaN-PIN, doping introduces an additional temperature dependence. In GaN, only a small fraction of Mg acceptors is ionized at room temperature. Therefore, rising temperatures lead to a higher density of free carriers by ionizing more impurity atoms. The presence of the electric field at the PIN junction assists this ionization process by Poole–Frenkel emission lowering the effective ionization energy of impurities. Such change of the free carrier density causes an optical modulation by BMS + BGR and affects the width of the depletion region. Furthermore, sample AlGaN-PIN is thicker than sample 2DEG-HS. This leads to a larger absolute change of optical path by temperature variation resulting in a higher sensitivity to temperature. Although, we did not measure the temperature behavior of sample GaN-PIN, we expect this sample to behave similarly since the total thickness and the doping concentrations are the same.

To avoid Ohmic heating through leakage current, the observed range was set to −8 V to 2 V for voltage-dependent reflectivity measurements. Figure 6 shows the relative reflectivity change due to an applied voltage \( \frac{\Delta R(V)}{R_0} = \frac{R(V) - R(0)}{R(0)} \). In these measurements, a truly electro-optical modulation can be proven by its polarity-dependent nature. For all samples, a strong modulation was present at \( \lambda = 362 \) nm near the bandgap of GaN and for sample 2DEG-HS additionally at \( \lambda = 325 \) nm near the bandgap of Al\(_{0.25}\)Ga\(_{0.75}\)N. Since Al\(_{0.25}\)Ga\(_{0.75}\)N is not covered by GaN in sample 2DEG-HS, short-wavelength features at the Al\(_{0.25}\)Ga\(_{0.75}\)N bandgap are not absorbed by GaN and can be detected. Additionally for longer wavelengths, an oscillating sub-bandgap modulation can be observed for all samples which is still present up to the detection limit of 820 nm.

Fig. 3. (Color online) CV graphs (black) and IV graphs (blue) of samples GaN-PIN, AlGaN-PIN, and 2DEG-HS. Graphs of samples GaN-PIN and AlGaN-PIN indicate a growing depletion region with stronger negative biases. Sample 2DEG-HS exhibits the characteristic step-like CV curve of a 2DEG.

Fig. 4. (Color online) Reflectivity spectra of the three devices.
the sub-bandgap region is caused by BMS+BGR which change the refractive index of GaN resulting in an absorption-free shift of Fabry–Pérot resonances for sub-bandgap wavelengths.\textsuperscript{15,21)} The cause of BMS+BGR is the modulation of the charge carrier densities by modulating the depletion width within the structures. Because the 2DEG in sample 2DEG-HS is negligibly thin compared to the doped p- and n-type layers in samples GaN-PIN and AlGaN-PIN, this effect is almost absent in sample 2DEG-HS. While the reflectivity changes continuously with growing reverse bias for the PIN samples, for sample 2DEG-HS, there is no further change in reflectivity when a bias more negative than \(-6 \text{ V}\) is applied (see Fig. 7). This correlates well with the complete depletion of the 2DEG.

4. Conclusion

We produced different active structures which are capable of a maximum electro-optical modulation of about \(\pm 1.5\%\) from \(-8\) to \(2\) V. FKE and BMS+BGR were identified as cause of the modulation; Ohmic heating was excluded as the root cause in the investigated voltage range. At this stage, it is not clear which device structure is the most promising; GaN-PIN and AlGaN-PIN show similar results, and the surface morphologies of the PIN samples and 2DEG-HS differ too much to permit a quantitative comparison. While the PIN samples show a broad sub-bandgap modulation, the 2DEG-HS offers a potentially higher switching speed due to the high electron mobility in the 2DEG. The low transparency of the semi-transparent electrode and the rough surface of p-GaN reduce the modulation signal significantly and will be optimized in the future. To amplify the observed modulation to technically applicable levels, future work will focus on integrating these active structures into a DBR.

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We attribute the features at the bandgaps of the respective materials to the FKE, distorting the absorption edge under the influence of an electric field. When a voltage is applied, the potential drops mainly across the i-layers of the three samples. Additionally for samples GaN-PIN and AlGaN-PIN, CV measurements have proven that the electric field is modulated in the doped GaN layers by shifting the limits of the depletion region.\textsuperscript{11)}

The modulation at the GaN bandgap of sample 2DEG-HS is much lower. Here, the electric field in GaN changes only in the thin region of the 2DEG quantum well.\textsuperscript{13–15)} The modulation in

![Fig. 6. (Color online) Voltage-dependent relative change of reflectivity of samples GaN-PIN, AlGaN-PIN, and 2DEG-HS normalized to the reflectivity at 0 V and room temperature absorption edge of GaN (dashed line) and Al\(_{0.25}\)Ga\(_{0.75}\)N (dotted line).](image)

![Fig. 7. (Color online) Relative change in reflectivity of sample 2DEG-HS near the bandgaps of GaN and AlGaN.](image)
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