INVESTIGATION AND CHARACTERIZATION OF AlGaN/GaN DEVICE
STRUCTURES AND THE EFFECTS OF MATERIAL DEFECTS AND PROCESSING
ON DEVICE PERFORMANCE

DISSERTATION

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The III-Nitride material system has proven extremely valuable for semiconductor device applications. The ability to grow high quality AlGaN/GaN that can be used for RF device applications is largely due to the commercial success of the implementation of p-type doping in GaN for optical devices. Even high quality GaN has relatively large defect densities. GaN devices are still able to achieve impressive performance, but not consistently. The variation in material quality, including deep-level defects and non-uniformities introduced by processing and growth, have deleterious effects on microwave device performance. These variations and the inability to control them reduce yield and reliability thus making AlGaN/GaN devices difficult to produce commercially. The purpose of this work is to characterize and contribute to the understanding of defects in AlGaN/GaN device systems and their effects on microwave device performance both DC and RF. The effects of device fabrication and surface processing on these defects have also been characterized.

Low Energy Electron-Excited Nano-luminescence (LEEN) Spectroscopy has been used to characterize radiative defects in the AlGaN/GaN material system on a microscopic scale and compare them with electrical measurements on HEMT’s and TLM structures. Salient features commonly observed in the LEEN spectra include donor-bound excitons in GaN at ~3.43 eV, donor-acceptor pair transitions (DAP) at ~3.30 eV,
yellow luminescence (YL) centered at ~2.20 eV, AlGaN donor-bound exciton emission, and associated phonon replicas. These measurements have been used to successfully correlate contact and sheet resistance with DAP, YL, and AlGaN near-band edge emission spectral features within a given wafer and between wafers. The effects of ultra-high vacuum processing with Argon sputtering and rapid thermal annealing on defects observed with LEEN spectra have been documented. Microscopic LEEN analysis has also been performed on working microwave devices and correlated to electrical measurements of frequency response, gain, and gate capacitance. Spectroscopic studies of working and failed microwave devices show that the surface and device processing changes have significant effects on device performance. These results show that it is possible to characterize and predict device performance in terms of deep level defects with non-destructive luminescence techniques on a very localized scale.
Dedicated to myself.
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LIST OF ABBREVIATIONS AND SYMBOLS

2-DEG – Two Dimensional Electron Gas
2-DHG – Two Dimensional Hole Gas
ADP – Auger Depth Profiling
AES – Auger Electron Spectroscopy
AFRL – Air Force Research Laboratory
BOE – Buffered Oxide Etch
CASINO – monte CARlo Simulation of electron trajectory in solids; program
CCD – Charge Coupled Device; used for optical detection
CFOM – Combined Figure Of Merit; compares a semiconductor’s performance to silicon
CL - CathodoLuminescence
CLS – CathodoLuminescence Spectroscopy
CMA – Cylindrical Mirror Analyzer
CRADA – Cooperative Research And Development Agreement
C-V – Capacitance Voltage; used to measure charge profiles
CVD – Chemical Vapor Deposition
DAP – Donor-Acceptor Pair; used to describe transitions from a donor to acceptor
DC – Direct Current
DI – De-Ionized water
DLTS – Deep Level Transient Spectroscopy

$E_A$ – Acceptor energy level

$E_b$ – Acceptor energy level

$E_C$ – Conduction band energy level

$E_D$ – Donor energy level

$E_F$ – Fermi level energy

$E_g$ – Bandgap energy

EHP – Electron-Hole Pair

$E_T$ – Trap energy level

$E_V$ – Valence band energy level

$\phi_b$ – Schottky barrier height; generic barrier height from conduction band to Fermi level

FET – Field Effect Transistor

FIB – Focused Ion Beam

$f_{MAX}$ – Unity power gain frequency

Freon-14 – Carbon tetrachloride used for isotropic RIE

Freon-23 – Carbon trifluoride used for anisotropic RIE

$f_T$ – Unity current gain frequency

FWHM – Full-Width Half-Maximum; description of spectral peak width

$g_m$ – transconductance

$h_{21}$ – $h$-parameter representing current gain

HBT – Heterojunction Bipolar Transistor

HEMT – High Electron Mobility Transistor
HFET – Heterojunction Field Effect Transistor

$h$-parameters – Hybrid parameters; microwave network characterization matrix parameters

HSA – HemiSpherical Analyzer

ICP – Inductively Coupled Plasma; used for anisotropic etching of nitrides

$I_{DS}$ – Current from drain to source at DC

$I_{dss}$ – DC saturation current from drain to source

ISA – Optical components company; Jobin Yvon Inc.

ISI – Insulator Seal; UHV component company

I-V – Current-voltage measurements usually referred to at DC

JEOL – Surface science equipment company specializing in microscopy

LEED – Low Energy Electron Diffraction; used to determine film crystal quality

LEEN – Low Energy Electron-excited Nanoluminescence; depth dependent CLS

LEO – Lateral Epitaxial Overgrowth

LO – Longitudinal Optical; a type of phonon

$L_T$ – Transfer Length; describes the effective ohmic contact length that carries current

MBE – Molecular Beam Epitaxy

MOCVD – MetalOrganic Chemical Vapor Deposition

MODFET – MOdulation Doped Field Effect Transistor

MSG – Maximum Stable Gain

$N_A$ – Acceptor concentration, not ionized

NBE – Near-Band Edge; used to describe photons with energy close to the bandgap

$N_D$ – Donor concentration, not ionized

ORIEL – Photonic components and instruments company

OSU – Ohio State University

PHI – Physical Electronics; UHV surface analysis equipment company

PL – PhotoLuminescence

PLS – PhotoLuminescence Spectroscopy

PMGI – PolyMethylGlutarImide; type of resist for multilayer lithography

PR - PhotoResist

$R_B$ – Electron penetration range in a material for an electron injected at a specific energy

RBD – RBD Enterprises; UHV surface analysis equipment company

$\rho_c$ – Specific contact resistance

$R_c$ – Contact resistance not normalized

RF – Radio Frequency

RIE – Reactive Ion Etch

$\rho_s$ – Sheet Resistance

RTA – Rapid Thermal Anneal

SEM – Scanning Electron Microscope

SI – Semi-Insulating

SMA – A standard threaded fiber optic connection type

SNR – Signal to Noise Ratio

$s$-parameters – Scattering parameters; microwave network characterization parameters

TLM – Transfer Length Method

UHV – Ultra-High Vacuum

UID – UnIntentionally Doped
UV – Ultra-Violet

$V_{DS}$ – Drain to source voltage at DC

WPAFB – Wright Patterson Air Force Base

Ψ – Electronic wave function related to electron positional dependence probability

YL – Yellow Luminescence
CHAPTER 1

INTRODUCTION AND BACKGROUND

1.1 Introduction

1.1.1 High Power High Frequency HEMT’s on AlGaN/GaN

Until recently, AlGaN/GaN material quality has been insufficient to produce high quality microwave devices. The recent improvements in GaN material quality that allow the fabrication of high frequency transistors have largely been driven by the commercialization of GaN for optical applications. These applications were made possible by the implementation of p-type doping in GaN by Amano et al. and the improved p-type doping concentrations by Nakamura in 1992.\textsuperscript{1,2,3} The original applications for the Group III-Nitride material system were laser diodes, light emitters, and UV detectors. Due to the direct, wide bandgap of these materials, these devices have been implemented in a wide range of color applications from the UV to visible including displays by incorporating varying amounts of Al, In, and Ga with a large degree of commercial success.\textsuperscript{3} More recently, AlGaN/GaN has been used for HEMT’s in high-frequency, high-power devices. Taking advantage of the piezoelectric charge forming a two-dimensional electron gas (2-DEG) at the AlGaN/GaN interface, many researchers have been able to implement HEMT’s with channel current-carrier densities an order of
magnitude higher than those formed with AlGaAs/GaAs. The high temperature stability of GaN compounds and high breakdown voltage of these materials allow very high density power operations which greatly exceed the power handling abilities of all other material systems to date with the exception of SiC. While HBT’s would yield superior frequency response and power handling, their implementation has been largely unsuccessful due to the inability to make high enough quality p-type GaN and reliable p-type ohmic contacts.\textsuperscript{4,5,6} The p-type ohmic contacts being used for LED’s are not good enough for microwave devices. Ohmic contacts to n-type GaN have been highly successful.\textsuperscript{7,8,9,10,11} Most III-Nitride microwave device research is currently focused on HEMT’s, which require only n-type ohmic contacts.

1.1.2 State of the Art Transistors

Even though GaN and related compounds still have a large number of growth related defects compared to other materials, HEMT’s made from AlGaIn/GaN heterostructures have impressive performance. As early as 2001, with 0.12 $\mu$m gate devices, $f_T$ and $f_{MAX}$ as high as 101 and 150 GHz have been reported for AlGaIn/GaN HEMT’s on SiC substrates.\textsuperscript{12,13} Other researchers report very high power operation with saturated output powers as high as 10.7 W/mm at 10 GHz.\textsuperscript{14,15} There have been many successful attempts at making HEMT’s on various substrates for improved power dissipation. Microwave applications include mobile-phone base stations, non-mobile wireless systems such as fixed wireless access systems, wireless local loop systems, or local multipoint distribution systems, and military applications such as tracking systems,
communications, and frequency jammers. GaN based devices are touted as being the material of choice for high performance devices for use in harsh operating environments since they require less cooling.
1.2 Background

1.2.1 AlGaN/GaN Properties

GaN has many desirable properties for electronic applications. Aside from the numerous optical applications, GaN is an excellent material for high frequency, high power devices. Compared to more mature material systems like Si and GaAs, GaN has already shown to be a formidable material. Since GaN has such a wide bandgap, it can be operated at higher temperatures and still remain a semiconductor. The wide bandgap also gives it a high breakdown field, permitting the scaling of devices an order of magnitude smaller than conventional materials as well as creating large conduction band offsets when growing AlGaN/GaN structures. This permits large sheet carrier densities to form at the AlGaN/GaN interface, which allows a higher gain in HEMT structures. GaN has a low thermal conductivity, but thinning the substrates or growing the structures on SiC easily solves this problem. Table 1 shows a comparison of several material parameters for Si, GaAs, SiC, and GaN.\textsuperscript{17} GaN has the highest overall performance figure of merit for microwave applications. The one disadvantage of the wide bandgap of GaN is the difficulty in making p-type GaN. N-type GaN can easily be made by Si doping with $N_D$ as high as $10^{19}$ cm\textsuperscript{-3} because the donor level is $\sim$ 30 meV below the conduction band. However, $N_A$ higher than $10^{18}$ cm\textsuperscript{-3} has not been demonstrated because the Mg acceptor level is $\sim$ 160 – 230 meV above the valence band causing only 1/100 of the acceptors to be activated.\textsuperscript{18} This makes it very difficult to fabricate HBT’s because of the high doping needed in the base, hence the focus on HEMT’s.
<table>
<thead>
<tr>
<th>Property</th>
<th>Si</th>
<th>GaAs</th>
<th>4H-SiC</th>
<th>GaN</th>
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<td>Bandgap $E_g$ (eV) @ 300 K</td>
<td>1.12</td>
<td>1.42</td>
<td>3.25</td>
<td>3.40</td>
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<td>Breakdown field $E_B$ (MV/cm)</td>
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<td>Electron mobility $\mu$ (cm$^2$/V s) @ 300 K</td>
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<td>8500</td>
<td>1000</td>
<td>1400</td>
</tr>
<tr>
<td>Maximum velocity $v_e$ ($10^7$ cm/s)</td>
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<td>2.0</td>
<td>2.0</td>
<td>3.0</td>
</tr>
<tr>
<td>Thermal conductivity $\xi$ (W/cm K)</td>
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<td>0.5</td>
<td>4.9</td>
<td>1.3</td>
</tr>
<tr>
<td>Dielectric constant $\varepsilon$</td>
<td>11.8</td>
<td>12.8</td>
<td>9.7</td>
<td>9.0</td>
</tr>
<tr>
<td>Sheet carrier density $N_{smax}$ (10$^{12}$ cm$^2$)</td>
<td>2.6</td>
<td>3.5</td>
<td>-</td>
<td>20</td>
</tr>
<tr>
<td>CFOM = $\xi\mu E_B^2 / (\xi\mu E_B^2)_{Si}$</td>
<td>1</td>
<td>8</td>
<td>458</td>
<td>489</td>
</tr>
</tbody>
</table>

Table 1.1. Comparison of common room temperature semiconductor parameters with combined figure of merit (CFOM) for high temperature, high power, and high frequency performance.\textsuperscript{17}

1.2.2 Issues with Device Fabrication

Sapphire is the most common substrate used for AlGaN/GaN growth. Due to the high lattice mismatch with GaN (13%) and poor thermal properties of sapphire, alternative substrates such as Si, SiC, AlN, GaN, GaAs, LiGaO$_2$ and ZnO have been implemented.\textsuperscript{19,20,21,22,23} There is no one substrate that is universally agreed upon, and different applications are better suited on different substrates. Also, different growth techniques are employed to make AlGaN/GaN, such as MBE, LEO, and variations of CVD.\textsuperscript{24,25,26} Hence, there is a great variety in the quality and types of problems
encountered for each device. However, one of the major obstacles that plagues the nitride device community in their efforts to produce commercial, high power, high frequency devices is repeatability and reliability. Within any given wafer, there is a very large difference in material parameters. Device parameters such as specific contact resistance, Schottky barrier height, sheet resistance, defect densities, layer thickness, Al mole fraction, and many others show drastic changes from die to die on a given wafer, not to mention from wafer to wafer.

Ohmic contacts can limit the transconductance of a device and are a major source of power waste with heat building up in the contacts. Schottky contacts are a source of leakage current, which can cause the gate to be destroyed with the application of higher voltages as well as preventing pinch-off. The contacts are generally where the device fails first assuming the material itself is of high enough quality to not degrade during operation. Surface states and bulk states in the AlGaN layer can contribute to current collapse in the devices as well as degradation of RF performance. It is believed that surface states on the AlGaN HEMT’s are the major contributor to current collapse and that passivation can reduce the effects of these states. However, it cannot remove them all. There has been a great deal of work to characterize the effects of device fabrication on device performance. Since there is no good wet-etch available for GaN materials, ICP etching and other physical etches are commonly used. ICP etching causes surface damage which is known to directly affect the properties of ohmic and Schottky contacts.

Even though there has been a great deal of work to characterize various etching parameters and process fabrication effects on electrical device performance, there has
been little done to understand the physical nature of these effects. Device engineers claim surface states and bulk defects cause current collapse and failing contacts, but have no physical explanation for what defects are present or what can be done to remove these defects. On the other hand, material scientists know a great deal about what defects are present in AlGaN/GaN materials and have a reasonable understanding of the energy levels of particular defects in GaN. However, they do not know what device fabrication procedures, if any, induce or reduce these defects or how the defects affect DC and RF performance. For successful RF device fabrication, it is essential to understand physically why these variations in material and device parameters exist and what can be done to control them in order to improve yield. Until very recently, it was unheard of to achieve microwave device parameter variations as small as 10% over a 2-inch wafer for AlGaN/GaN. Even on very good wafers, variation in contact resistance is as high as ~15% while variations in sheet resistance can be as low as 1%.

1.3 Research Outline

In this work, a correlation is made between defects found in the AlGaN/GaN device structures and electrical measurements performed on the devices. The first experiments involve the characterization of AlGaN/GaN HBT’s with cross-sectional cathodoluminescence spectroscopy to extract depth dependent defect information from the samples of interest. The difference in spectral features is noted. The initial experiments suggest that there is a correlation between spectral features associated with acceptors in GaN and average n-type ohmic contact resistance values obtained for different samples. These contaminating acceptors were from the segregation of Mg
doping during the growth. The change in contact resistance being associated with compensation due to unintentional doping begs the question of how do unintentional dopants from unknown elements, defects intrinsic to the material, or chemical processing steps affect the actual device performance.

After the initial spectroscopic measurements, a set of carefully controlled experiments was used to correlate spectral features associated with defects and measurements of sheet and contact resistance on AlGaN/GaN HEMT layers with a thick GaN cap. The samples had identical structure with the exception of Al mole fraction. Spatially localized LEEN spectra were taken in an SEM at each TLM site used to characterize the material and contact quality. The same area on the TLM structure was observed during device fabrication and the effects of annealing the samples in various ambient gases was observed spectrally. The deep-level defects were observed to increase and decrease as a function of processing and the yellow luminescence was found to correlate with the contact resistance but not the sheet resistance.

The spectral features related to the AlGaN correlate well with sheet resistance. Deep level defects or Al mole fraction changes in the AlGaN were shown to account for the variations observed in the sheet resistance. The defects were found to exhibit localization at interfaces and the surface of the GaN cap. The fact that the Al mole fraction did not appear to affect the contact resistance in these samples indicates that the defects at the surface of the sample or in the cap layer itself were responsible for the electrical variations observed and not intrinsic properties of the material. The material used for this experiment exhibited large changes in contact properties and thus changes within a wafer and between wafers were readily observed.
Since DC contact and material parameters were shown to be related to the defects found in the AlGaN/GaN material, the microwave response of HEMT’s should change dramatically based on what defects might be found in the material. Complete individual HEMT devices from two wafers provided by a commercial vendor were studied with LEEN and AES. The only known properties of these devices are the device structure and that one wafer produced devices with RF results and the other produced no working devices. Samples from the different wafers exhibit different spectral signatures. AES maps of the devices reveal the presence of Cl in a device suffering catastrophic failure. The few devices studied reveal that the spectral signatures related to the quality of the material could be correlated to acceptable and failing high-frequency response. Knowing this, a set of experiments was planned to study the effects of defects observed spectroscopically at the gate region of a given device and how these defects respond to surface treatments.

AlGaN/GaN HEMT material was acquired from another commercial source for surface treatment and device fabrication studies by observing the LEEN spectra at the gate region of individual devices. A new ultra-high vacuum compatible gate mask was created so the gate metal could be deposited under very clean conditions. The plan for this sample set was to use identical ohmic contacts while varying only the gate processing under controlled UHV conditions. A control sample using the standard gate deposition was compared with samples that were sputtered gently with Ar and annealed in vacuum prior to gate deposition. The spectral features were to be compared with the microwave device performance. The effects of ultra-high vacuum processing were observed spectroscopically. Unfortunately, at the final stages of gate lift-off, the samples were
irreversibly destroyed prior to performing any RF measurements. However, correlations between device fabrication, DC measurements, and spectral features related to donor-acceptor pairs and yellow luminescence have been found. Modifications to the surface of the gate region in UHV also result in changes in the LEEN spectral signatures indicating a change in the defect structure.

In the absence of RF measurements from a HEMT sample set that was fabricated and characterized at OSU, a comparison between RF measurements and spectral features observed on individual devices was made on state of the art AlGaN/GaN HEMT’s on SiC substrates prepared off-site. Electrical measurements of the devices repeat the correlations observed on previous samples relating donor-acceptor pair and yellow luminescence to contact resistance. The sample also exhibits trends in gain and frequency response radially across the wafer that are correlated to defect spectral features. These results suggest that the changes in defect spectra due to processing on earlier samples would impact RF performance. It has been shown that variations in device parameters can be predicted and characterized by their spectral features. The nature of the role of the defects in the III-Nitride system can be explained providing insight regarding the contributions of intrinsic material defects, processing induced defects, and geometry effects to device performance. These results will show the value of spectroscopic characterization of deep level defects in AlGaN/GaN as a tool that can analyze and predict frequency response for individual HEMT’s. These experiments will be used as feedback for optimizing device fabrication to create superior HEMT’s.
CHAPTER 2

EXPERIMENTAL TECHNIQUES

2.1 Ultra-High Vacuum

Ultra-high vacuum (UHV) is a condition in which controlled experiments can be performed without introducing any contaminants into the system. Standard device fabrication processing is performed under atmospheric conditions in clean rooms with low particle counts. Class 100 cleanrooms are common, which means the number of airborne contaminants greater than 0.5 \( \mu \text{m} \) in diameter is less than 100 \( \text{ft}^{-3} \). While humidity and temperature can be controlled very well in these facilities, there are airborne contaminants and simple compounds in the atmosphere that cannot be filtered out and do affect device performance, namely oxygen, hydrogen, water vapor, and organic compounds. In the UHV chambers used in these experiments, base pressures routinely drop as low as \( 6 \times 10^{-11} \) Torr where 760 Torr is atmospheric pressure. There are so few particles in the chamber at any given time that it would take 24 hours for a monolayer of adsorbates to form on the surface of the sample. In these ultra-high vacuum chambers, device samples can be annealed, metalized, and analyzed without extraneous impurities affecting the outcome of the measurements. By altering the fabrication procedures of the gate metalization in vacuum, there is no doubt which
parameter has been altered to induce the change in device performance for a given material. In these chambers, the capacity to perform LEED, CLS, and AES measurements in real time allow meaningful physical parameters to be extracted while defects and impurities form or reduce during the processing of the GaN device layers.

2.2 Cathodoluminescence Spectroscopy

Cathodoluminescence spectroscopy (CLS) is the resolution of the energies of photons produced by an electron excitation source irradiating the sample of interest. There are many reasons for using CLS as a characterization tool for semiconductors. One of the main advantages is evident when the number of electron-hole pairs (EHP’s) generated in the material per incident electron is considered. Typically, the number of EHP’s generated is dependent upon the ionization energy \( E_i \) of the material, which is given by \( E_i = 2.8 E_g + M \).\textsuperscript{34} The bandgap of the material is \( E_g \) and \( M \) is a material dependent parameter with a magnitude of about 0 – 1 eV. This means on the order of \( E_b/E_i \) EHP’s are generated where \( E_b \) is the beam energy and has typical values of 1 – 40 keV. The large number of EHP’s generated allows many electronic states, which would be very difficult to detect otherwise, to be generously excited producing a strong luminescence signal. Fortunately, this technique can be used at energies up to 100 keV without producing atomic displacement.\textsuperscript{35} Although in some instances, CLS is known to produce changes in the material system at low beam energies (< 5 keV).\textsuperscript{36}

Using CLS, defects and material parameters from AlGaN/GaN device layers can be observed. The most prominent emissions are typically from the near band edge emission (NBE) of GaN and Al\(_x\)Ga\(_{1-x}\)N. At room temperature, GaN NBE is observed at
3.43 eV and Al\textsubscript{x}Ga\textsubscript{1-x}N NBE is observed higher than GaN with a typical value for x = 0.15 being ~3.75 eV. Monitoring AlGaN NBE can be a quick way to gauge mole fraction and sheet resistances in HEMT’s. Other common features observed at room temperature are yellow luminescence (YL) centered at 2.20 eV. Donor-acceptor transitions can be observed at ~3.25 eV and other deep level defects can be seen from ~2.8 – 3.2 eV. The presence of these deep level defects can be detrimental to device performance and can also be used as an indicator of potential device performance prior to fabrication.

2.3 LEEN Spectroscopy

The most powerful aspect of CLS is the penetration depth dependence on $E_b$. Varying the incident beam energy allows the material of interest to be probed for electronic information in a non-destructive fashion as a function of depth in the sample.\textsuperscript{37,38,39,40,41,42,43} When CLS is performed at varying beam energies in this fashion, the technique is referred to as Low Energy Electron-Excited Nano-Luminescence (LEEN) spectroscopy to emphasize the surface localization of the beam. At energies less than 1 keV, the electron penetration depth can be as small as tens of angstroms.\textsuperscript{34,44} Today, surface science routinely uses only monolayers of material deposition at a time to make device structures such as quantum wells and HEMT cap layers that can be as thin as 2 nm. A basic understanding of the electron beam penetration properties is required. The first analytical approach for modeling the electron beam depth dependence by Everhart and Hoff and a Monte Carlo approach are discussed for predicting the electron energy distribution in samples for the injected beam.
A fairly simple model of electron energy loss as a function of depth is given by Everhart and Hoff. This model can be used with fair accuracy for $1 \text{ keV} < E_b < 25 \text{ keV}$ for low atomic numbers. The Everhart-Hoff model begins to break down at low energies due to its empirical nature making it difficult to apply to very shallow structures near the surface. Everhart and Hoff begin by citing a general energy loss per path length, given by:

$$
-\frac{\partial E}{\partial s} = \left(2\pi N_A e^4 \frac{Z \rho}{A} \right) \left[ \frac{1}{E} \ln \left( \frac{a E}{I} \right) \right]
$$

(2.1)

where $e$ is the electronic charge, $N_A$ is Avogadro’s number, $Z$ and $A$ are the atomic number and gram atomic weight, $\rho$ is the material density, $I$ is the mean excitation energy with a dependence on $Z$, $E$ is the electron energy, and $a$ is a constant equal to 1.1658.

The theoretical maximum range for an incident electron, $R_B$, is then given by:

$$
R_B = \int_0^E \frac{\partial E}{\partial (\rho s)} = \int_0^{(E_b)} \frac{\xi (\rho s)}{\ln (\xi)}
$$

(2.2)

where a variable substitution, $\xi = aE/I$, has been made for ease of notation. This becomes important when low energies are considered. The factor $\xi$ determines which empirical range formulas can be used since at low energies the integral goes to infinity as $\xi$ approaches 1. Everhart and Hoff determine an empirical value for $R_B$, which is given by Equation 2.3.

$$
R_B = \left( \frac{0.0398}{\rho} \right) E^{1.75} (\mu\text{m})
$$

(2.3)
This range is calculated by taking into account material parameters such as atomic number and density for a SiO\textsubscript{2} layer. It has been observed that the dependence of the energy term on the exponent in Equation 2.3 is slightly different for different incident energy ranges due to the dependence of $\xi$ on energy and atomic number. This particular range is best for $5 \text{ keV} < E < 25 \text{ keV}$ and for $10 < Z < 15$ due to the value of $\xi$ for a SiO\textsubscript{2} layer.

The depth dose curve is a measure of the electron energy loss distribution in the solid as a function of position. Everhart and Hoff use a third order polynomial with empirically determined coefficients to model the depth dose curve. The polynomial is given by $g(z) = 0.60 + 6.21z - 12.40z^2 + 5.69z^3$, for $0 < z < 1$, and $z$ is normalized to the range, $R_b$. Figure 2.1 shows the Everhart-Hoff predictions for the electron distribution from 0.5 – 5.0 keV beams. The theory predicts very shallow penetration depths, even at higher energies. The penetration depth for a 0.5 keV electron beam is calculated to be on the order of 30 Å. The Everhart-Hoff predictions are attractive due to the analytical form of the solution, which can be easily used for calculations and quick estimations of the electron beam depth dependence.
More recent theories are based on Monte Carlo statistical electron trajectory calculations. These calculations simulate anywhere from $10^2$-10$^7$ electron trajectories. The programs take into account material density, which is strongly related to the scattering cross-section of the material, and the energy loss per collision at a given point. The computer randomly generates a number that determines which angle the electron will be scattered. The simulation then subtracts energy from the incident electron until its kinetic energy is zero and the position is recorded. Using a larger number of simulations increases the accuracy of the Monte Carlo calculations, but there is a trade off between computation time and desired accuracy. The CASINO simulator was designed by SEM users with a thorough understanding of electron-solid interaction and chosen because it was readily available freeware with the capabilities to model multiple junctions.\textsuperscript{46,47,48,49}
The results seem reasonable and correlate well with physical observations in this research. Figure 2.2 shows the results of using CASINO to model the depth dependence. From Figure 2.2, the Monte Carlo theory predicts very shallow penetration depths as well. The penetration depth for a 0.5 keV electron beam is simulated below to be on the order of 80 Å. Comparing Figure 2.1 to Figure 2.2 shows that the Everhart-Hoff calculations seem to underestimate the depth dependence when compared to new Monte Carlo models. CASINO results are undoubtedly more accurate, but they are not in an analytical form and require a computer to be interpreted. For careful sample analysis Monte Carlo simulation is the desired approach, but if a general beam location is desired, the analytical approach yields reasonable qualitative information as well as some insight into the physical processes.

Figure 2.2. Depth does curves for CASINO electron beam penetration Monte Carlo simulations as a function of beam voltage at constant power.
2.4 Auger Electron Spectroscopy

Auger Electron Spectroscopy (AES) is the spectroscopy of core electron energies for particular atoms. AES involves a three-electron process in which a core electron is removed by an injected electron beam. An electron in a shell with higher energy can make a transition into the hole that is created. To conserve energy, an X-ray photon is released or another electron can be released. An electron released in this process is known as an Auger electron and has an energy level that is characteristic of the core energy levels of a particular atom. The kinetic energy of electrons released in this process is measured with a Cylindrical Mirror Analyzer (CMA) and recorded electronically. The signal strength can then be used as a measure of elemental composition in the sample of interest. AES is a surface-sensitive technique primarily because of the escape depth of the electrons that are measured. Typical Auger transitions are observed lower than 1600 eV. As a result, Auger emission can only be observed from ~30 Å or less. AES can be used to detect elemental compositions as low as 1 – 2 %. So, while it is useless to detect impurities at dopant concentrations, it can be used to see if the surface of the sample has residual carbon or oxygen layers. This is a useful technique for UHV surface preparation prior to metalization.

AES can also be used in conjunction with inert ion sputtering to perform Auger Depth Profiling (ADP). Typically argon is used as an inert gas source to remove layers of the sample at a slow rate (Å/s). By sputtering the sample of interest and performing AES on the target spot, the elemental composition as a function of depth can be resolved. This is particularly useful when trying to analyze multi-layer structures. In this case, observing the movement of multi-layer ohmic and Schottky contacts after annealing can
be helpful in determining how a contact is formed or perhaps what element is degrading a contact. It can also be used to find gross contaminants within a given layer of the contact.

AES energy levels are also sensitive to the surrounding material matrix. These levels can shift because of mole fraction, strain, and impurities. Even monitoring the shift of Auger energy levels can provide clues that there is change occurring in the device and contact layers.

2.5 Scanning Electron Microscope

A JEOL 7800F Scanning Electron Microscope (SEM) is available for sample analysis. SEM’s create sample images by recording the intensity of secondary electrons emitted as a primary electron beam is rastered over the area of interest. The minimum observable feature size is limited by the primary electron beam diameter. In the case of the JEOL 7800F, the best resolution obtainable is ~5 nm under ideal conditions with a beam voltage of 25 kV. The secondary electrons emitted are also sensitive to the electronic properties of the surface of the sample providing additional information about surface states.

This particular SEM has been retrofitted with an optical train that allows LEEN spectroscopy to be performed by using the primary electron beam as the optical excitation source. Optical emissions are collected using an Oxford MonoCL apparatus consisting of a parabolic collection mirror on the UHV side coupled through a quartz viewport to an air side monochromator and photomultiplier tube detection system. The beam voltage on the SEM can be adjusted from 0.1 – 25.0 kV with adjustable currents.
ranging from $5 \times 10^{-11} \text{ – } 1 \times 10^{-8}$ A. The large range of beam energies available allows a large range of depths to be probed with CLS from as shallow as 20 Å to as deep as 2.5 µm. Panchromatic and monochromatic light images can be taken of samples for luminescence mapping. This particular optical train can be used in conjunction with a cold stage capable of reaching temperatures as low as 10 K with liquid He cooling.

Additionally, this SEM has a hemispherical analyzer (HSA) that detects Auger electrons generated by the primary electron beam. Using the AES detection abilities combined with the mapping abilities of this instrument, elemental compositions across the surface of a sample can be observed for device structures. This microscope is a powerful tool for measuring device materials and actual devices. Spatially localized information about where defects occur in the material as well as how they correlate with electrically measured properties gives valuable feedback about contact formation as well as material growth. The resolution of the SEM permits the investigation of luminescence properties and elemental compositions in areas between device contacts on finished devices that are difficult to analyze by conventional means. Figure 2.3 shows a schematic of the SEM fitted with the light collection and AES equipment along with a typical LEEN or CLS spectra for AlGaN.
Figure 2.3. Schematic of the JEOL 7800F SEM fitted with the Oxford MonoCL and HSA. The sample is mounted under the light pipe that collects the optical emissions from the electron beam. A typical AlGaN spectrum is shown as the output of the system on the left.

2.6 Electron Beam and Thermal Evaporation of Metals

In order to prepare clean metal-semiconductor interfaces for ohmic and Schottky contacts to AlGaN/GaN, it is necessary to deposit the contacts in a controlled environment. The standard method of contact formation is electron-beam evaporation. Normally, electron-beam evaporation is performed by placing a clean metal source in a ceramic crucible in a vacuum chamber. An electron beam is focused onto the metal source until it evaporates onto the semiconductor. The thickness of the film is measured by monitoring the change in resonant frequency of a crystal oscillator in the vacuum chamber. Under normal device fabrication conditions, the evaporations usually take place at $\sim 1 \times 10^{-6}$ Torr. At these pressures, there are enough molecules in the chamber to
form a monolayer of adsorbate every second. Any contaminants in the chamber during evaporation can stick to the sample. Due to the high temperatures of the sources, which are exposed to atmospheric conditions every time a sample is loaded, the potential for outgassing contaminants into the chamber during evaporation is high.

Another method of metal deposition is thermal evaporation. In this method, the clean metal source is placed in a crucible wrapped with a tungsten filament. Current is passed through the filament until the crucible heats and the metal evaporates. The metal thickness is monitored again with a crystal oscillator. This method can be used for metals with lower melting points such as Au, Al, and Ga. With careful outgassing procedures, thermal evaporators can be operated at $5 \times 10^{-9}$ Torr or better. Thermal evaporation is a very clean method of contact deposition, but does not work well for refractory metals such as Pt and Ti, which are common metals used in GaN contacts.

2.7 Current Voltage Measurements

To extract useful parameters from the contact made to the AlGaN/GaN device layers, simple I-V measurements are made on test structures patterned on the HEMT layers. The simplest measurements that can be performed are on Transfer Length Method (TLM) structures. The current and voltage are measured on pads of varying spacing and the resistance is plotted as a function of pad spacing. As the limit of the pad spacing approaches zero, the non-zero resistance that is observed is attributed to the specific contact resistance ($\rho_c$). Sheet resistance ($\rho_s$) and transfer length ($L_T$) can also be extracted from these measurements. Figure 2.4 shows a sample TLM plot along with how to extract the parameters. For a HEMT structure, $\rho_s$ is a measure of the formation of
the 2-DEG at the AlGaN/GaN interface. Contact resistance is the resistance to current flow at the ohmic contact. The transfer length represents how far laterally under the ohmic contact the current will flow and is related to $\rho_c$ and $\rho_s$. Other simple measurements include I-V curves of the Schottky diodes formed at the gate, from which the barrier height ($\Phi_B$) can be extrapolated. The barrier height measurements can give information regarding surface states at the gate metal-semiconductor junction.

Figure 2.4. TLM plot of resistance vs. contact spacing showing the extraction of various contact parameters.
In addition to contact information, DC I-V curves of the drain current as a function of the drain-source voltage for different applied gate biases can give information regarding transconductance \((g_m)\) and channel saturation current \((I_{dss})\).\(^5\) RF measurements, which are more complicated to make, require network analyzers and high quality equipment. The microwave performance is obtained by measuring \(s\)-parameters, \(h\)-parameters, unitary short-circuit current gain frequency \((f_t)\), maximum frequency of oscillation \((f_{max})\), maximum stable gain (MSG), and others.\(^6\) The devices can behave very differently at microwave frequencies, which are dependent on traps in the bulk and at the interfaces and surface of the devices.\(^7\) Passivation of the devices can help minimize the effects of surface traps, but does not eliminate the effects completely.\(^8\) Hence, one of the focuses of this study is the effect of gate processing on AlGaN/GaN HEMT RF performance.
3.1 Auger Depth Profiling

An ADP system has been constructed in a UHV chamber with a base pressure of $6 \times 10^{-11}$ Torr using a single pass PHI 15-110A CMA and PHI electronic boxes 20-085, 11-045A, 11-065, 20-805, and 20-075. A PHI 04-303 ion gun with Ar and N gas sources has been implemented and aligned to the focal point of the CMA for depth profiling. The instruments are controlled and monitored by an RBD 147 controller and RBD data acquisition card in a computer. The sputter gun is at an oblique angle to reduce sputtering rates and to prevent elements from being pushed into the sample by ion bombardment. Sputtering rates ranging from 4 – 500 Å/min have been measured on metal and semiconductor samples. The primary use of this system is surface preparation for ohmic and Schottky contacts. The AES portion is used at low voltages to detect surface cleanliness and the ADP system is used to gently clean contaminants, which cannot be removed by flash annealing in vacuum. Damage from the sputtering process can be reduced by annealing the sample after sputtering.
3.2 Fiber-Optic, Ultra-High Vacuum, Cathodoluminescence Spectroscopy

In the same UHV chamber, an in-situ, fiber-optic CLS assembly has been designed and constructed for LEEN spectroscopy. A PHI 04-015A electron gun with a PHI 11-010 variable power supply is used to probe the sample with 0.1 – 5.0 keV electron beams. This allows electron beam penetration depths from 2 – 170 nm for excellent depth resolution. Using a 600 µm core fiber with transmission from 180 – 1200 nm in conjunction with a UHV-SMA feedthrough flange from ISI, the fiber can be brought to within a millimeter of the sample or closer for maximum light collection efficiency compared to other lens systems. The light from the fiber is coupled from the external flange on the chamber to an ORIEL MS260i monochromator by an ORIEL F/# matcher. The monochromator has a 2400 l/mm holographic grating for excellent spectral resolution from 180 – 600 nm and a 600 l/mm grating for lower resolution spectra from 250 – 1300 nm. The light is collected by an Instaspec IV CCD system. The CCD allows rapid data collection as well as flexibility in taking measurements by adjusting exposure times for low signal samples. The total calibrated system response is from 180 – 900 nm or 6.88 – 1.38 eV, which allows every possible combination of Al$_x$Ga$_{1-x}$N/GaN to be studied as well as quaternary compounds including In. The NBE of AlN is observed at 6.2 eV and the NBE of InN should be observed at 1.9 eV. The fiber is covered at all times by a stainless steel shield when not in use. Using this system in the same chamber as the AES/ADP system and metal evaporation components, the AlGaN/GaN gate layers can be monitored for surface and sub-surface defects with LEEN spectroscopy before and after annealing, sputtering, and metalization steps.
3.3 Metal Evaporation Sources for Ti, Pt, Al, Au, and Ga in Ultra-High Vacuum

Common ohmic contact schemes for n-type GaN are Ti/Al/(Ti, Ni, Pt)/Au layers.\textsuperscript{59,60} Common Schottky contacts for n-type GaN are Ni/Au, Pt, and Pt/Au.\textsuperscript{60,61,62} In order to put down contacts, Ti, Pt, Al, and Au sources have been implemented. Ti and Pt are refractory metals and are difficult to evaporate thermally. A dual-source rod-fed electron beam evaporator from Thermionics has been implemented for Ti and Pt. In order to maintain good chamber pressure during evaporation, water-cooling lines have been implemented on the source flange. Since Al and Au are easier to evaporate, thermal evaporators were built from boron nitride and alumina crucibles wrapped with tungsten filaments. The current is supplied through a 150 A feedthrough by a Sorenson DCS20-150 power supply. Film thickness from 100 – 3000 Å has been demonstrated for all materials. The thickness is measured with a crystal oscillator and QXM500 thickness monitor from Kurt J. Lesker Company.

A thermally evaporated Ga source has been implemented for surface processing on GaN. One technique for obtaining LEED ordered surfaces on GaN involves evaporating monolayers of Ga onto the surface of GaN and annealing in UHV. Preparing a clean surface on the AlGaN/GaN HEMT layers prior to gate deposition will be instrumental in comparing surface preparations with RF device performance. Differential pumping lines at the source flanges have been implemented to further reduce chamber pressure during metalization. So far, pressures during evaporation range from $1.1 \times 10^{-8}$ – $6.3 \times 10^{-9}$ Torr for electron beam evaporation and thermal evaporation respectively. These pressures are significantly lower than the standard device fabrication pressures and
ensure that extrinsic contaminants from the metal depositions are reduced. Shielding has also been built to control the deposition of the metals and protect the analysis equipment.

3.4 CRADA Development at AFRL for Device Fabrication Facilities

AlGaN/GaN materials require aggressive dry etching techniques to perform mesa isolations. State of the art device processing and measurement for AlGaN/GaN HEMT’s also requires other equipment that is currently unavailable at The Ohio State University. In order to acquire the necessary resources to perform the work of interest, CRADA 00-SN-02-W was formed between the Sensors Directorate at Wright Patterson Air Force Base (WPAFB) and Professor L. Brillson at The Ohio State University. Through this CRADA, the university has the ability to perform ICP etching for mesa formation along with automated device measurement systems, stress testing and RF measurement equipment. The university also benefits by receiving MBE GaN material grown at WPAFB for fabrication and characterization. In return, WPAFB receives material characterization and feedback from the device fabrication for their material growers. Additional arrangements are being made to secure an industrial source of AlGaN/GaN HEMT structures through this CRADA for the final phases of this research project.
As LEEN Spectroscopy is the most commonly used characterization tool for AlGaN/GaN device structures in this work, it is helpful to become familiar with commonly observed radiative recombination mechanisms found in AlGaN/GaN cathodoluminescence spectra and their physical significance. In this chapter, the importance of band-to-band recombination, intentional doping, and unintentional doping will be discussed. Other optical features such as excitonic and phonon related transitions will be explained and spectra with examples of some of these features will be shown. Red-shifts in luminescence spectra will also be explained in terms of band bending and the Franz-Keldysh effect.

Since radiative recombination is due to electron and hole interaction with states in the band gap of the material, it is necessary to understand the band structure and associated defects in the AlGaN/GaN system. More work to date has been done to characterize defect position in GaN than AlGaN and most features resolved in this work are related to the GaN layer itself. Thus, the focus of the recombination mechanisms will be discussed as they relate to the GaN layer and analogies to AlGaN will be made with the assumption that the Al$_x$Ga$_{1-x}$N for smaller values of Al content
0.30>x behaves similarly to GaN. In either event, the AlGaN features observed in this work are related to near-bandedge recombination and occasionally its associated phonon replicas. The band structure noting common radiative defects is shown for GaN in Figure 4.1. Several LEEN spectral features related to these transitions are shown in Figure 4.2. These figures will be referred to in the course of the chapter to explain the features observed in the following sections. The recombination mechanisms discussed in this section are by no means considered complete nor do they represent non-radiative transitions that are present in the system.

Figure 4.1. GaN band structure showing radiative recombination paths at room temperature. Electrons are excited from the valence band to the conduction band and return through various recombination paths.
4.1 Band-to-Band Recombination Related Luminescence

The simplest and most likely radiative recombination path in a direct bandgap semiconductor like GaN is the band-to-band recombination. In this transition, the excited electron returns directly from the conduction band to the valence band emitting a photon with energy equal to the bandgap. At room temperature, this is about 3.425 eV. At low temperature, 11 K, it is 3.474 eV. However, in practice, the direct band-to-band recombination is never observed directly. Donor-bound and free excitonic transitions are commonly observed instead.\textsuperscript{67,68,69}

Excitons are an electron and hole bound weakly by coulombic interaction. The potential energy from this weak interaction results in a shift in the energy required to make a transition to and from a given state. This is seen as a peak at a lower energy than
expected for the transition without any excitonic influence. Donor bound excitons and free excitons are all commonly observed as one peak at low resolution or room temperature. For ease of reference, the term near-band edge (NBE) emission is used to loosely describe all of the unresolved transitions that appear at or near the band edge energy in the optical spectra. The NBE related transition is noted in Figure 4.1. Examples of NBE spectral features for both n-type AlGaN (3.962 eV) and n-type GaN (3.474 eV) are shown in Figure 4.2.

4.2 Intentional Dopant Related Luminescence

The common elements used for doping GaN are Si for n-type and Mg for p-type. Activation energies for silicon dopants have been reported as low as 22 meV.70,71 Due to the optical limitations involved in these spectra, either the resolution was too poor to resolve the transition from using large slits to accommodate low power measurements or the measurements were taken at room temperature. Most of the time, in these measurements, the donor to valence band transition is not resolved and included in the NBE measurements. This transition can sometimes be observed as a small shoulder on the low energy side of the NBE emission.

Activation energies for the magnesium dopants have been reported from 150 – 250 meV.70,72 Transitions from the donor levels into the acceptor level are referred to as donor-acceptor pair (DAP) transitions and can sometimes be resolved at room temperature at ~3.1 – 3.25 eV. However, the radiative recombination efficiency is very low at room temperature, but the features are readily observed for p-type material. They
are easily observed at 10 K. The DAP transition is observed at 3.281 eV in unintentionally doped (UID) n-type GaN at 10 K in Figure 4.2.

4.3 Unintentional Dopants and Defect Related Luminescence

Many of the AlGaN and GaN layers in these studies are unintentionally doped. However, even the best UID GaN has donor concentrations of $\sim 2 \times 10^{16} - 5 \times 10^{16}$ cm$^{-3}$. The unintentional shallow level donors come from oxygen substitutional impurities in a nitrogen site, nitrogen vacancies, and background silicon contamination. The largest contributor to the UID doping concentration is probably oxygen. These shallow donors do not manifest themselves optically in any special way. They are also absorbed into the NBE emissions and not resolved.

In many n-type and UID GaN films, DAP transitions can be observed. As demonstrated in the GaN band structure in Figure 4.1, the DAP in UID films is a result of background magnesium contamination in the growth reactors or possibly carbon. Magnesium is far more probably since it has been used in most reactors at some point in an attempt to make p-type GaN films. Figure 4.2 shows how the DAP transitions manifest themselves in the low temperature spectra of UID GaN films at 3.281 eV. These transitions do not have a high radiative recombination efficiency. Thus it is necessary to cool the sample to observe the transition clearly.

The most commonly observed defect in all n-type GaN luminescence studies is the yellow luminescence (YL). Even now, the nature of the YL is not completely known. However, it is generally accepted that this feature is the result of a shallow donor to a deep level acceptor. The shallow donors are any of the oxygen or silicon in a nitrogen
site or nitrogen vacancies. The deep acceptor is believed to be 1.0 – 1.2 eV above the
valence band. This acceptor is thought to be a combination of various defects, namely
gallium vacancies and gallium vacancy complexes with oxygen and hydrogen.\textsuperscript{74,75,76,77,78,79} This spectral feature is observed in almost every n-type GaN
sample to some degree. It is easily observed at room temperature and does not appear to
increase much at low temperature as seen in Figure 4.2.

4.4 Phonon Replicas

Phonons are the quantized lattice vibrations in the crystal. Electrons can absorb
or create phonons when making optical transitions within the bandgap. In high quality
crystals, phonon related transitions are common. Optical (LO) phonons have an energy
of \(~90\) meV. Phonon replicas of features in the spectra appear as peaks in multiples of
the quantized 90 meV energy level below the main feature energy. Figure 4.2 shows
phonon replicas of the AlGaN and GaN NBE at 3.875 and 3.388 eV and two LO phonons
of the DAP spectral features at 3.188 and 3.100 eV. In some samples, phonon replicas
have been observed in the yellow luminescence.\textsuperscript{77}

4.5 Quantum Effects at AlGaN/GaN Interfaces

Due to the size of the features under investigation, it is not uncommon to observe
red-shifted NBE features in either AlGaN or GaN. The intense band bending due to the
piezoelectric charge at the AlGaN/GaN interface is the mechanism behind the operation
of the AlGaN/GaN HEMT’s. This intense band bending also gives rise to tunneling
assisted recombination from the conduction band to the valence band in the AlGaN and
GaN also known as the Franz-Keldysh effect.\textsuperscript{80,81} Figure 4.3 shows a Franz-Keldysh transition for a quantum well. In this quantum well under an applied electric field, $E$, the bands bend such that the electronic wave function penetrates into the forbidden region of the smaller bandgap semiconductor and the electron can recombine into the valence band state by emitting a photon of lower energy than the bandgap plus the energy levels of the wells. This effect does not require a quantum well to work, only extreme band bending. Red-shifts in both AlGaN and GaN spectra are observed due to this effect. These are the main features and mechanisms observed in the spectra during the course of this work.

![Franz-Keldysh effect diagram](image)

Figure 4.3. Franz-Keldysh effect shown for a quantum well. The electron wave function penetrates into the forbidden region because of the band bending in the presence of the electric field, $E$. The electron is able to recombine at a lower energy than when the bands are flat.
CHAPTER 5

AlGaN/GaN HBT N-TYPE OHMIC CONTACTS: A DC TEST CASE

5.1 Ohmic Contact Optimization and Electrical Measurements

In 2000, MBE grown HBT structures from AFRL and a commercial source, Company A, were processed with the intent of optimizing n-type ohmic contacts to the emitter and collector and p-type ohmic contacts to the base. While attempts at making p-type ohmic contacts resembled leaky Schottky contacts, the n-type ohmic contacts performed well. Ohmic contacts to n-GaN were prepared with various metalization schemes and processing conditions. The HBT’s had the following nominal layer structure: GaN(200-nm, \(N_D = 1.8 \times 10^{19} \text{ cm}^{-3}\))/GaN-Al\(_{20}\)Ga\(_{80}\)N(20-nm, \(N_D = 1.8 \times 10^{19} \text{ cm}^{-3}\))/Al\(_{20}\)Ga\(_{80}\)N(80-nm, \(N_D = 5.0 \times 10^{17} \text{ cm}^{-3}\))/Al\(_{20}\)Ga\(_{80}\)N(5-nm, UID)/GaN(150-nm, \(N_A = p^+\))/GaN(500-nm, \(N_D = 3.0 \times 10^{16} \text{ cm}^{-3}\))/GaN(1400-nm, \(N_D = 1.0 \times 10^{18} \text{ cm}^{-3}\))/Al\(_2\)O\(_3\). The layer structure is graphically depicted in Figure 5.1.
Using Ti/Al/Ti/Au and Ti/Al/Pt/Au metal layers and varying anneal temperatures from 800 – 1000 °C and \( N_D \) from 3.0x10^{16} – 2.0x10^{19} cm\(^{-3}\), the optimal contact resistance was found to be 3.0x10^{-6} \( \Omega \)-cm\(^2\) for \( N_D = 2.0x10^{19} \) cm\(^{-3}\) with Ti(10-nm)/Al(100-nm)/Ti(50-nm)/Au(20-nm) contacts annealed at 850 °C for 30 seconds in a nitrogen ambient. The material was cleaned with a dilute 10:1 DI:NH\(_4\)OH solution prior to contact deposition. Metalization took place in an electron-beam evaporator at \( \sim 1.0x10^{-7} \) Torr. The contacts were treated with a rapid thermal anneal (RTA) in a nitrogen gas ambient. Contact resistance measurements were performed at AFRL on transfer length method (TLM) structures as discussed in Section 2.7.

Figure 5.2 shows a summary of all the ohmic contact values obtained. Each run shown is the result of averaging ohmic contact resistances over \( \sim 15 \) die. AFRL material consistently produces ohmic contacts with lower resistance than Company A. Runs 1 – 6 are contacts formed on AFRL material and 7 – 10 are contacts to Company A material.
The \( E \) and \( C \) notation refers to ohmic contacts on the emitter and collector respectively. Since the emitter doping is three orders of magnitude higher than the collector, the emitter ohmic contacts are generally better than the collector. Runs 1E, 2E, and 3E consist of Ti(10-nm)/Al(100-nm)/Ti(50-nm)/Au(20-nm) contacts annealed at 800 °C for 30 seconds in a nitrogen ambient. Runs 1C, 2C, and 3C consist of Ti(10-nm)/Al(100-nm)/Pt(50-nm)/Au(20-nm) contacts annealed at 800 °C for 30 seconds in a nitrogen ambient. Runs 4E, 5E, and 6E are the same as Run 3E except annealed at 850 °C, 900 °C, and 1000 °C respectively. Runs 4C, 5C, and 6C are also the same as Run 3C except annealed at 850 °C, 900 °C, and 1000 °C respectively. Run 7 is the same as Run 1E. Run 8 is the same as Run 1C. Run 9 consists of Ti(10-nm)/Al(100-nm)/Pt(50-nm)/Au(20-nm) contacts annealed at 850 °C for 30 seconds in a nitrogen ambient. Run 10 consists of Ti(10-nm)/Al(100-nm)/Pt(50-nm)/Ni(20-nm) contacts annealed at 800 °C for 30 seconds in a nitrogen ambient. The average ohmic contact resistance for all AFRL HBT samples is \( 7.5 \times 10^{-6} \, \Omega \text{-cm}^2 \) while the Company A average is \( 1.1 \times 10^{-4} \, \Omega \text{-cm}^2 \).
Figure 5.2. Summary of ohmic contact resistance obtained for each run on AFRL and Company A HBT material. Each run is the average value obtained over ~15 die for each piece of material. The E and C notation refers to ohmic contacts on the emitter and collector respectively. AFRL material consistently produces lower resistance ohmic contacts than Company A material.
5.2 Cross-Sectional CLS Characterization

The samples were then analyzed in cross-section at OSU by CLS with 25 kV electron energies and 1 nA beam current in the JEOL JAMP 7800 SEM on the Oxford low temperature stage at 11 K as in Section 2.5. Light collection was achieved in the usual way except in this instance the signal was recorded with a Triax320 monochromator and CCD-3000 detection system from ISA. Figure 5.3 is included to show the relative scales of the HBT structure. The HBT layers are shown in the figure on the left side as indicated. The dark band in the middle of the HBT layers is the $p^+$ GaN base region. The beam was focused at 300,000X and scanned across the HBT layers. Spectra of the emitter, base, and collector region were obtained.

Figure 5.3. Cross-sectional SEM image of AFRL HBT. The left side of the image is the surface. The HBT layers are indicated. The dark band in the middle of the HBT layers is the $p^+$ GaN base region. The GaN/Al$_2$O$_3$ interface is clearly indicated. The beam was positioned at high magnification to obtain CL spectra in the various layers.
There are several salient features that are common in both sample sets. AlGaN and GaN near-band edge (NBE) emission are observed as well as an acceptor-related deep level near 3.3 eV associated with the donor-acceptor pair (DAP) transition from a shallow donor to a Mg acceptor. Figure 5.4 shows the spectra taken from the surface of the sample into the bulk sapphire for an AFRL HBT. The AlGaN NBE emission is measured at 3.70 eV while the GaN NBE emission is measured at 3.52 eV. The DAP luminescence is measured at 3.23 eV and relatively small. As expected from the nominal growth structure, the AlGaN features are predominantly located just under the surface in the emitter layer and reduce in intensity beyond the base layer. The feature at 3.23 eV is visible at the surface, but reaches a maximum intensity deeper into the sample than the maximum intensity of the AlGaN maximum intensity. All features vanish as the beam moves beyond 1900 nm into the sapphire substrate. Significant yellow luminescence (YL) at 2.20 eV is observed in the surface GaN cap layer on this sample.
Figure 5.4. AFRL HBT cross-sectional CLS spectra from surface to substrate. Salient features include AlGaN and GaN band edge emission, Mg acceptor emission, and yellow luminescence.

Figure 5.5 shows the same CL spectra from a Company A HBT. The features are similar with a few differences. The AlGaN and GaN NBE energy levels are measured at 3.88 eV and 3.50 eV respectively. The DAP feature has a peak intensity at 3.31 eV which is closer to the valence band than AFRL HBT’s. The DAP luminescence is also much more prominent and extends further into the sample than the AFRL samples. The acceptor luminescence also exhibits phonon replicas that indicate higher quality crystal structure. There is also no YL observed at the surface of this sample.
Figure 5.5. Company A HBT cross-sectional CLS spectra from surface to substrate. Salient features include AlGaN and GaN band edge emission, Mg acceptor emission, and YL. The sharp step at 3.1 eV is an artifact of the grating spectra being pieced together in the CCD data acquisition software.

Figure 5.6 shows a direct comparison of CL spectra in the emitter and collector region for both HBT samples. The top two spectra are from Company A material and the bottom two spectra are from AFRL material. The emitter and collector spectra are clearly labeled. The main differences to note are from sample to sample. Company A material has much higher integrated deep level emission intensity at 3.32 eV than AFRL material which is at 3.24 eV. The Company A material shows phonon replicas of the
DAP luminescence and the AFRL material does not. The variations from emitter to collector spectra are insignificant with the exception of the slight low energy shoulder on the GaN NBE emission in the higher doped emitter of the AFRL HBT. This is likely due to the increase in donors.

Figure 5.6. Comparison of CL spectra in emitter and collector regions for AFRL and Company A HBT’s. The Mg related luminescence at 3.24 – 3.32 eV is much stronger in the Company A sample set. The AFRL samples exhibit a slight broadening on the low energy side of the GaN band edge with higher n-type doping.
5.3 Data Correlation and Discussion

Low temperature CLS studies at 11K show major spectral differences between AFRL and Company A HBT’s. The largest difference in the two sample sets is the characteristic shape of the DAP luminescence. In these samples, this acceptor luminescence is attributed to the Mg source in the growth chamber that is intentionally introduced to dope the base layer. When the Mg acceptor level is observed in other layers, it is attributed to segregation effects during growth. In Company A samples, the Mg luminescence is a dominant feature suggesting high levels of electrically active acceptors. It is also found throughout all of the HBT layers in Company A samples suggesting either very heavy contamination or the diffusion of Mg from the intentionally doped base layer. AFRL HBT’s show less contamination or diffusion. However, both samples show some acceptor luminescence in layers where it would not be expected. Some of this could be due to electron diffusion and rastering from the analysis beam used to generate the optical spectra. This spreading is small though and only accounts for some of it. A more plausible explanation is that the Mg acceptors are compensating the Si donors in the collector and emitter layers. This explanation is validated by comparing the ohmic contact values between the two samples. Company A ohmic contact values are high and there is a large Mg acceptor luminescence observed when compared to the GaN band edge emission intensity. AFRL ohmic contacts are more than an order of magnitude lower and the Mg related luminescence is significantly lower. Furthermore, the acceptor level in Company A samples is closer to the GaN band edge, which indicates a shallower acceptor that is more readily ionized. AFRL samples have a deeper acceptor that would not ionize as well. The deeper level of the Mg acceptors in the AFRL samples
suggests very high level concentrations of Mg atoms as reported elsewhere during
colorization of p-type GaN by PLS.\textsuperscript{82,83,84}

The AlGaN NBE energy levels are higher in Company A samples indicating a
higher Al mole fraction. It might be argued that the higher Al content creates a wider
bandgap material that would result in higher ohmic contact values near the emitter.
However, when forming ohmic contacts to the collector, the entire AlGaN layer is etched
away and the emitter ohmic contacts follow the same trend as the collector ohmic
contacts. So, it appears that the Al content has no effect on these measurements.

This experiment has demonstrated that spectral features from CLS on pieces of
bulk material can be correlated with average ohmic contact values from real device
structures. The effect of expected residual impurities that are present in the growth
chamber has been observed spectroscopically and compared with electrical
measurements. This cross-sectional technique while yielding useful information about
the device layers is not appropriate for all sample sets. It is not always prudent to
repeatedly cleave an expensive sample on a sapphire substrate until an appropriate
surface is found for spectral analysis. There is not a Focused Ion Beam (FIB) available
with CLS functionality, and sawing the wafer would destroy any chance at observing the
interfaces. There is also no way to look at individual devices with cleaving and thus this
analysis must be compared with average values assuming the spectral features are
comparable for all areas of the wafer. Since it has been demonstrated that these electrical
and optical measurements can be correlated with average values, these experiments are
now performed on individual devices on HEMT layers taking advantage of the non-
destructive, spatially localized, depth dependence of LEEN Spectroscopy with the SEM
as in Section 2.3. The effects of the same defects on ohmic contacts will be investigated on a sample set which should nominally have no Mg segregation because all of the layers are grown n-type or unintentionally doped.
CHAPTER 6

AlGaN/GaN A400’S HEMT LAYER OHMIC CONTACT INVESTIGATION

6.1 AlGaN/GaN Sample Preparation

From January - April 2001, the HEMT layers used in this experiment were grown on sapphire by MBE at AFRL. Samples A401, A418, A424, and A425 have the following structure: GaN (50-nm, UID)/Al\(_x\)Ga\(_{1-x}\)N (67.5-nm, UID)/GaN buffer (UID)/SI-GaN (1-\(\mu\)m)/AlN nucleation layer/Al\(_2\)O\(_3\). The SI GaN layer was formed by Mg compensation doping. Figure 6.1 shows a schematic of the layer structure. Each sample is a quarter of a 2-inch wafer and contains about 15 die. A401 and A418 have nominal mole fractions of \(x = 0.15\), and A424 and A425 have nominal mole fractions of \(x = 0.10\). All samples have a buffer thickness of 500 nm except for A418, which has a 250 nm thick buffer. Mesa isolation was performed by ICP dry etching. The ohmic contact layers consisted of Ti (10-nm)/ Al (100-nm)/ Ni (50-nm)/ Au (20-nm) annealed at 800 °C for 30 seconds in a nitrogen ambient. All of the wafers were processed at the same time to eliminate any possible deviations related to the contact preparation. TLM measurements provided values for contact resistance (\(R_c\)), specific contact resistance (\(\rho_c\)), transfer length (\(L_T\)), and sheet resistance (\(\rho_s\)). The contact spacings were later verified using an optical microscope with a calibrated CCD imaging system. Figure 6.2 shows a
schematic diagram of the fabrication process and at which points LEEN spectra were taken. The purpose of this experiment is to correlate ohmic contact performance with defect emission spectra at the measured TLM site.

Figure 6.1. Layer structure for A400’s series HEMT’s grown at AFRL.
Figure 6.2. Graphical process follower for HEMT fabrication. The circled numbers 1 – 3 indicate when LEEN spectra were taken between processing steps. The thin layer with hashed lines is the active HEMT layer made from AlGaN/GaN. The source, gate, and drain are indicated by S, G, and D respectively.


6.2 Ohmic Contact Electrical Measurements

Using the TLM analysis method as described in Section 2.7, the specific contact resistance and sheet resistance was extracted for ~15 die on each of the four quarter wafers used in this experiment. Figure 6.3 and Figure 6.4 show the statistical summary for the specific contact resistance and sheet resistance respectively. The boxes denote one standard deviation around the median value represented by the bar in the middle of the box. The whiskers on the boxes denote outliers and encompass 1 – 99% of all data points. There is a large spread spanning two orders of magnitude in the contact resistance in these samples. The values for contact resistance appear to vary regardless of the nominal Al mole fractions of these layers. The contact resistance range is from $2 \times 10^{-6}$ – $2 \times 10^{-4}$ $\Omega$–cm$^2$ over all samples. The sheet resistance is well defined for all samples except sample A418. The sheet resistance does behave as expected by increasing for samples with nominally lower Al mole fraction and decreasing for samples with nominally higher Al mole fraction. The range in sheet resistance measured is from 500 – 2400 $\Omega$/sq.
Figure 6.3. Specific contact resistance summary for the four sample wafers obtained by TLM. The boxes represent one standard deviation in the value above and below the median value dividing the box. The whiskers on the boxes denote the outliers and 1 – 99% of all values obtained. The dot inside the box denotes the mean value.

Figure 6.4. Sheet resistance summary for the four sample wafers obtained by TLM. The boxes represent one standard deviation in the value above and below the median value dividing the box. The whiskers on the boxes denote the outliers and 1 – 99% of all values obtained. The dot inside the box denotes the mean value.
6.3 Ohmic Contact LEEN Characterization

6.3.1 Measurement Description

As described in Section 2.5, the quartered wafers were analyzed by LEEN Spectroscopy in the JEOL 7800F SEM during processing. The electron beam spot size used was 150 nm rastered over a 3.5 \( \mu \text{m} \times 3.5 \mu \text{m} \) square area. The spectra were taken at room temperature since the low temperature stage was unavailable at the time. Using the SEM, LEEN spectra were taken in the middle of the largest contact spacing of the TLM structure on many die of each quarter wafer with electron beam energies ranging from 1.0 – 4.0 keV. The depth of the electron beam penetration varied from 17 – 130 nm, as calculated from CASINO. Monte Carlo simulations of penetration depth versus beam energy are shown in Figure 6.5 for these particular structures. At 1.0 keV and 2.0 keV, the beam penetration is confined to the top GaN cap layer. At 3.0 keV, the beam penetration extends to the first GaN/AlGaN interface, while at 4.0 keV, the luminescence spectra have nearly equal contributions from the GaN cap layer and the AlGaN layer. It is important to note that CASINO does not account for carrier diffusion after generation. Hence, it is possible to see features with LEEN that are slightly deeper than CASINO predicts. Figure 6.6 shows the expected luminescence contribution by region as found from integrating the CASINO curves in the AlGaN and GaN regions. The spectra of different TLM structures are obtained for many die as a function of beam energy. Optical emissions were collected in the 1.75 – 4.25 eV photon energy range and all emissions are normalized to the GaN NBE emission. The beam currents and energies were adjusted to low power levels to reduce or eliminate time dependence in the spectra. For all normalized spectra, the YL is integrated in the 1.75 – 2.70 eV range. This integrated
value, as well as other deep level emissions, will be compared with contact resistance and sheet resistance values from die to die at different beam energies. Figure 6.7 shows a schematic of the LEEN technique applied to the TLM structures. Using the SEM, individual device structures can be analyzed and the analysis can be repeated at the same location between processing steps. The figure also shows a typical spectra obtained from the A400 sample set.

![Figure 6.5](image)

Figure 6.5. CASINO simulation showing electron beam penetration for the beam energies used in the A400 sample set. The sample surface starts at 0 nm and the bulk is along the positive x-axis. The GaN/AlGaN/GaN layers are clearly noted.
Figure 6.6. Expected luminescence contribution for GaN and AlGaN as a function of beam energy as derived from integrating CASINO simulations.

Figure 6.7. Schematic showing LEEN analysis of the TLM structures. The beam is injected into the middle of the largest TLM spacing and the photons are collected resulting in the spectra on the left. The photo shows a blow-up view of the TLM from one of the die.
6.3.2 Annealing and Low Temperature Fabrication Effects on LEEN Spectra

A lot of work has been done by other researchers to characterize the effects of etching on AlGaN/GaN device properties, especially the rectifying properties of the gate under various etching conditions. However, the effects of low temperature and annealing fabrication steps on HEMT layers are relatively unknown. The TLM structures on these samples have been monitored after various fabrication steps. LEEN spectra have been taken of the unprocessed material, after ohmic contact annealing, and after gate metalization as indicated in Figure 6.2.

Spectra of the material during fabrication can be found in Figure 6.8 and Figure 6.9. Figure 6.8 is a set of LEEN spectra taken from A401, die(6,11). The spectra are plotted in order of increasing beam energy from bottom to top. There are three sets of curves for each beam energy. The solid curve is the unprocessed material. The dashed curve is directly after the ohmic anneal. The dotted curve is after the fabrication is complete. Figure 6.2 shows schematically the point in processing at which LEEN spectra were taken. Figure 6.9 is the same plot as Figure 6.8 except the measurements were performed on A425, die(4,10). The main differences in these spectra are the dramatically larger YL features observed in the unprocessed A401 spectra in Figure 6.8. Wafer A401 also has a larger average contact resistance than A425. In both cases, the YL is dramatically reduced by the ohmic contact anneal at 800 °C for 30 seconds in nitrogen. The YL is again reduced by the processing performed for the gate metalization. The only thermal process involved for this is a bake at 270 °C for 5 minutes in air on a hot plate while covered with PMGI to cure the gate mask. The deep level luminescence at ~3.1 eV is also reduced continuously during fabrication. It is difficult to see differences in this
defect luminescence in unprocessed material from sample to sample. The observation of this feature is further complicated by the room temperature analysis. The spectra taken from the ohmic anneal to the gate metalization were taken at the exact same location. Therefore, there is no question that any changes in the spectral features are from the processing and not from location dependent variation in the wafer properties.

Figure 6.8. LEEN characterization of A401 die (6,11) during fabrication. From bottom to top, the spectra are from the surface to the bulk regions respectively. The solid curves indicate unprocessed material; step(1). The dashed curves indicate spectra taken after the ohmic anneal; step(2). The dotted curves indicated spectra taken after fabrication is complete; step(3). Yellow luminescence and deep level luminescence decreases during processing.
Figure 6.9. LEEN characterization of A425 die (4,10) during fabrication. From bottom to top, the spectra are from the surface to the bulk regions respectively. The solid curves indicate unprocessed material; step(1). The dashed curves indicate spectra taken after the ohmic anneal; step(2). The dotted curves indicated spectra taken after fabrication is complete; step(3). Yellow luminescence and deep level luminescence decreases during processing.

With the depth dependence of LEEN, the spectra show AlGaN NBE emission intensity increasing from 1.0 – 4.0 keV. The peak just begins to appear at 2.0 keV, which is expected from CASINO simulations shown in Figure 6.5 allowing for diffusion. In Figure 6.8 showing spectra of sample A401, the YL appears to strongly peak in intensity at 3.0 keV, which corresponds to the beam penetrating the top GaN/AlGaN interface. This phenomenon was observed on several die suggesting that some of the YL is associated with GaN/AlGaN interface related defects. However, after fabrication was
complete, this apparent localization of the YL at the interface was not observed suggesting that the interface was at least partially repaired after processing.

6.3.3 Low Temperature Induced Defects

To determine whether the low temperature effects seen in Figure 6.8 and Figure 6.9 were from the low temperature PMGI curing step instead of induced by the gate metalization, pieces of the unprocessed material were annealed in air at 270 °C to simulate gate lithography curing steps. There was no PMGI cap on the material during this experiment as there would be during fabrication. Figure 6.10 shows the LEEN spectra at 4.0 kV of unprocessed material compared to material annealed in air. The solid curves are the unprocessed material. The dashed curves are from the annealed material. The YL does not appear to change appreciably, but there is an increase in the deep level defect emission intensity at 3.1 eV for all samples. The largest increase was observed on the samples with the largest average contact resistance and the material with the best contacts showed little increase. The samples are plotted from lowest average ohmic contact resistance to highest from top to bottom. It is uncertain whether oxygen or water vapor played a role in this transition or how it might affect luminescence attributed to Mg impurities or other deep level defects. It is possible that hydrogen was present in the material and the low temperature anneal was enough to break the passivating bonds to the acceptors. These broken bonds would allow the acceptor to be electrically activated and hence observed optically. These results show that low temperature processing can have effects on devices that are intended for use as high power operation where local temperatures could easily reach these fabrication temperatures.
Figure 6.10. LEEN spectra of unprocessed material at 4.0 kV compared to material annealed at 270 °C in air for all A400 samples. Deep level defect emission centered around 3.1 eV increases after the anneal showing the smallest increase on samples with lowest ohmic contact resistance. The plots are in order from lowest average ohmic contact resistance to highest looking from top to bottom.
6.3.4 Deep Level Defect Emission Compared to Contact and Sheet Resistance

Figure 6.11 shows the integral of the YL as a function of ohmic contact resistance. The plot contains the data from 2.0 – 4.0 keV. Signal-to-noise ratios at 1.0 keV were too low to perform calculations with acceptable accuracy. Ohmic contact resistance correlates well with increasing YL intensity, showing a statistically significant correlation. For a more accurate measurement, each TLM structure spacing was optically verified and the specific contact resistances were recalculated for the same die that were measured with LEEN. The data from A424 has been omitted because, unlike the other samples, it showed an unphysically lower contact resistance from the original measured values when rechecked for accuracy. In contrast, the contact resistance of the other samples changed by 10 - 15% during the correction.

Figure 6.11. Plot of YL integrated intensity normalized to GaN band edge as a function of specific contact resistance. The square series is for 2.0 kV beam voltage. The x shapes denote 3.0 kV beam voltage. The circle series denotes 4.0 kV beam voltage. In all cases, the amount of YL directly correlates with increasing contact resistance.
No correlation is observed between AlGaN NBE energy level and contact resistance. From Figure 6.10, the average contact resistance of a given wafer shows the NBE of the AlGaN is at higher energies for both the lowest and highest contact resistances. In a few cases, the AlGaN emission was observed at beam voltages lower than 3.0 keV, indicating a decrease in the distance between the metal-GaN and GaN/AlGaN interfaces. From LEEN spectra, the variations in layer thickness are 5 – 10%. If the contacts were prepared directly on AlGaN or a very thin GaN cap, the higher bandgap material could affect the metal-GaN barrier. In this case, a correlation between AlGaN NBE energy level and contact resistance might be expected.

Deep level defect emissions in the AlGaN layers have also been identified by LEEN as a signature of uncharacteristically high sheet resistances for a given wafer. For the same ohmic contacts in this chapter, average sheet resistance values ranging from 650 – 2275 Ω/sq as Al mole fraction decreases. The defect emissions associated with high sheet resistance are found to be in the AlGaN layer itself since they do not appear in spectra with e-beam voltages less than 3.0 keV. Figure 6.12 shows the sheet resistance plotted as a function of AlGaN NBE emission. The sheet resistance drops with increasing Al mole fraction. This is consistent with a higher piezoelectrically induced sheet charge density formed at the AlGaN/GaN interface from lattice mismatch induced strain. This shows that the 2DEG formed at the AlGaN/GaN interface as expected. There is an outlier in this plot from a die on A418, which is noted. The sheet resistance on this particular die is 1150 Ω/sq whereas all other die on the wafer average 630 Ω/sq. The LEEN spectra for this particular die shows strong deep level emissions from the AlGaN layer in the 3.0 and 4.0 keV spectra. Figure 6.13 shows a comparison between
the spectra at 4.0 keV of a TLM with a sheet resistance of 590 $\Omega$/sq and the TLM with extremely high sheet resistance of 1150 $\Omega$/sq. The overlap between the AlGaN and GaN peaks is evident. The spectra of the TLM with the lower sheet resistance is representative of all die on this wafer having low sheet resistance.

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**Figure 6.12.** Plot of sheet resistance as a function of AlGaN band edge energy level. The sheet resistance decreases as the AlGaN band edge energy level increases consistent with higher piezoelectrically induced sheet charge density from higher Al content and lattice mismatch induced strain.
Figure 6.13. LEEN spectra comparing the spectra of a representative die on A418 to a spectra of the die on A418 that has an abnormally high sheet resistance for that wafer. The solid curve represents the normal die and the dashed curve represents the abnormal die. The overlap between the AlGaN and GaN band edge emissions is clear in the abnormal curve suggesting an effectively lower Al mole fraction.

6.4 Discussion

Various concentrations of Mg doping are known to cause a range of deep level defects from 3.1 – 3.2 eV and 2.8 – 3.1 eV. It is possible that the Mg used to compensate the insulating GaN layer is present in residual quantities during the device layer growth. Sample A418 does not show any significant luminescence other than GaN NBE and YL, implying that these deep levels do not exist here or that they recombine non-radiatively. If these deep levels are associated with Mg, it is possible that the acceptor sites generated here could compensate some of the available donors at the metal-semiconductor junction. The lower intensity of these deep level emissions relative
to GaN NBE at higher beam voltages indicates that these defects are confined near the surface of the sample and not at the GaN/AlGaN interface. The interface defect luminescence should become more pronounced as thinner or no GaN cap layers are used on new devices.

The YL is believed to be associated with transitions from a shallow donor, e.g., a N vacancy or O in a N site, to a deep acceptor level, e.g., a Ga vacancy.\cite{75,76,77} In these samples, the YL is also more prominent at the surface of the sample as is commonly reported elsewhere.\cite{77,103} Yellow luminescence is an indicator of material quality, presumably increasing as material properties degrade. The deep level acceptor associated with the YL could also produce some compensation effects, albeit not as efficiently ionized as the acceptors seen at \(~3.1\) eV. However, the defects associated with this luminescence may also serve to pin the Fermi level at the metal-semiconductor interface closer to an intrinsic level.\cite{104} Therefore, both deep level defects and Mg impurities could cause ohmic contacts to be highly resistive by reducing the effective doping level.

Figure 6.11 provides some evidence of the surface localization of the YL because the 2.0 and 3.0 kV data show a stronger trend than the 4.0 kV data, which shows more scatter. This indicates that the contact resistance has a stronger dependence on the surface states than states found deeper in the bulk.

Decreasing sheet resistance associated with an increase in channel carrier density correlates well with increasing AlGaN NBE energy level which is attributed to the increasing Al mole fraction. Since there is no correlation between AlGaN NBE energy level and contact resistance, these observations are consistent with the role of defects near the metal-semiconductor interface and the large separation of this interface from the
AlGaN layer. Furthermore, anomalously high sheet resistances can be identified by defect spectra in the AlGaN layer. These defects found in the spectra between the AlGaN and GaN NBE can be attributed to variations in Al mole fraction at the affected site. No correlation is observed between sheet resistance and YL intensity.

Ohmic contacts to AlGaN/GaN device layers have been fabricated, measured, and characterized with LEEN. Examining unprocessed material with LEEN shows trends that can predict device performance prior to fabrication. Spectral information can be used to characterize differences in electronic performance from wafer to wafer and within a wafer from die to die. LEEN has also been used to characterize and predict the sheet resistance of a given wafer as well as identify regions of uncharacteristically high sheet resistances on a particular die. The existence of radiative deep level defects can be used to identify regions of poor performance. However, it is important to consider that the absence of radiative defects does not guarantee desirable device performance, as there may be non-radiative defects that inhibit performance. Using this combination of electrical measurements and spectral information, the nature of these defects can be understood and controlled to improve material quality and reliability. Unprocessed material shows that samples with the lowest amounts of midgap luminescence produce the lowest contact resistance. Experiments showing spectral changes during HEMT fabrication also give clues to reliability during high power operation.

In this chapter, the results indicate that the defects associated with the YL play a significant role in ohmic contact performance. It does indeed have a strong impact on the ohmic contacts by creating regions where the Fermi level is pinned locally, but later results will show that the defects associated with the YL are only a part of the full picture.
Since the low temperature stage of the SEM/CL equipment was unavailable during this experiment, the role of the acceptors at ~3.1 eV was not quantifiable in the same fashion as the YL defects because the luminescence signal was very close to the noise. In this sample set, it is expected that Mg impurities will contribute to the results, but the YL of these samples is very large compared to material from other vendors. Hence this case may be strongly affected by pinning at the YL defect level. In many samples, Mg is not used to compensate or create a SI layer. The effects of intrinsic acceptors and defects associated with AlGaN/GaN on device properties still need to be addressed. Additionally, since LEEN spectral correlations with electrical measurements show the effects of defects on contact performance, a natural extension of this experiment is to apply this technique to microwave devices. The next section describes the characterization of a set of AlGaN/GaN HFET’s by focusing on the gate region. These experiments anticipate an effect from deep level defects observed in the gates of these FET’s having an impact on RF device performance.
7.1 Company B, GS-15 and GS-17 Sample Set

The samples used in this experiment are individual transistors taken from two wafers GS-15 and GS-17. They both have the following structure: Al$_{0.25}$Ga$_{0.75}$N (10-nm, UID)/Al$_{0.25}$Ga$_{0.75}$N (10-nm, $N_D = 1.0 \times 10^{19} \text{-cm}^{-3}$)/Al$_{0.25}$Ga$_{0.75}$N (30-nm, UID)/GaN (3-$\mu$m)/SiC. Figure 7.1 shows a schematic of the HEMT layer structure. The samples were grown by MOCVD and the finished devices were passivated with 250 nm of silicon nitride. The devices are multi-finger with air bridges. Standard ohmic and Schottky contacts for GaN are used for these devices. All devices are processed identically. Average ohmic contact resistance was $2.5 \times 10^{-6} \Omega \cdot \text{cm}^2$, and average sheet resistance was 430 $\Omega$/sq. Complete electrical information from the supplier is not available. However, it is known that both sets of devices have identical DC characteristics. The RF measurements on FET’s from GS-15 had RF power and gain, while FET’s from GS-17 did not function at RF.
Figure 7.1. Layer structure for AlGaN/GaN HEMT’s grown on SiC for GS-15 and GS-17.

7.2 LEEN Spectral Comparison of Working and Inoperable Devices

As described in Section 2.5, the quartered wafers were analyzed by LEEN Spectroscopy in the JEOL 7800F SEM during processing. The electron beam spot size used was 50 nm rastered over a 1.1 μm x 1.1 μm square area. The spectra were taken at room temperature since the low temperature stage was still unavailable. Using the SEM, LEEN spectra were taken in the gate region of HFET’s between the source and gate, on the part of the mesa not covered by the air bridge. Figure 7.2 shows an SEM image of where the LEEN spectra were taken. The gate, source, and drain are clearly labeled. The LEEN analysis spot is indicated between the air bridge metal and mesa edge, and between the gate finger and source/drain. Electron beam energies of 3.0, 4.0, 5.0, 6.0, 8.0, and 10.0 keV were used. The depth of the electron beam penetration varied from 140 – 630 nm, as calculated from CASINO. Monte Carlo simulations of penetration depth versus beam energy are shown in Figure 7.3 for these particular structures. At 3.0 and 4.0 keV, the beam is confined to the silicon nitride passivation layer. At 5.0 keV the
beam just begins to penetrate the AlGaN barrier layer. At 6.0 and 8.0 keV, a large portion of the energy is deposited into the AlGaN barrier layer, while at 10.0 keV, most of the energy is deposited into the GaN bulk. Again, the CASINO simulations do not account for carrier diffusion once generated. Hence, it may be possible to see features with LEEN that are slightly deeper than CASINO predicts. The spectra for several transistors are observed for wafers GS-15 and GS-17. Again, beam currents and energies were adjusted to power levels low enough to prevent time dependent spectral changes due to localized heating and charging. General comparisons are made between the spectra in terms of working devices compared to inoperable devices.
Figure 7.2. SEM image showing LEEN analysis point on GS-15 and GS-17 HFET’s. The analysis spot is directly between the gate finger and source/drain and between the mesa edge and the air bridge.
Figure 7.3. CASINO simulation for GS-15 and GS-17 electron depth dependence for LEEN spectra. The surface of the sample is to the left with the silicon nitride passivation layer, AlGaN layers, and GaN bulk clearly marked. The beam does not significantly penetrate the AlGaN layer until after 4.0 kV.
From these samples, the expected behavior for the LEEN spectra is shown in Figure 7.4. This figure shows the integrated electron distribution by region for AlGaN and GaN. The integrated regions are from 250 – 300 nm and 300 – 3300 nm for AlGaN and GaN respectively. The luminescence associated with each region is expected to track qualitatively as the integrated values in the figure. This simplified estimate neglects carrier diffusion, quantum, field effects, and optical re-excitation (internal photoluminescence). From this estimate, AlGaN and GaN should have no contribution before 5.0 kV while AlGaN feature intensities should peak around 6.0 – 7.0 kV. The GaN features will slowly begin monotonically increasing at 5.0 – 6.0 kV and be completely dominant at higher voltages.

Figure 7.4. CASINO electron energy distribution integrated by region as a function of beam voltage. The GaN curve shows the CASINO curves integrated from 300 – 3300 nm. The AlGaN curve shows the curves integrated from 250 – 300 nm. These curves represent the expected relative luminescence contribution from each region.
Optical measurements of GS-15 and GS-17 produced very different results. The transistors measured from GS-15 that exhibit high frequency operation follow the theoretical expectation somewhat. Figure 7.5 is a set of LEEN spectra on GS-15 die (5,6) transistor number 61. This particular FET shows a gradual increase in GaN signal intensity that dominates the spectra at higher beam voltages while the AlGaN band edge signal intensity seems to peak up around 6.0 kV. The 6H-SiC feature at around 2.9 eV begins to show up clearly at higher beam voltages due to secondary photoluminescence, which is excitation from the primary photons emitted from recombination of the electron-hole pairs generated by the electron beam. YL is also observed at 2.20 eV. In this particular transistor, it appears localized at 5.0 kV. The most interesting features in these spectra are the evolution of the GaN NBE emission from low beam voltages to high. At 3.0 – 4.0 kV beams, the GaN NBE emission appears at 3.26 eV. At voltages greater than 4.0 kV, the GaN NBE shifts to 3.41 eV. This apparent red shift in energy of 150 meV near the surface of the sample can be attributed to a Franz-Keldysh effect at the triangular quantum well found at the AlGaN/GaN interface from the piezoelectric band bending effect responsible for the 2-DEG.
Figure 7.5. LEEN spectra of a working HFET number 61 from die (5,5) on GS-15. The AlGaN band edge luminescence contribution is clearly visible as is a red-shifted GaN band edge feature near the top of the FET structure. Spectra at higher energies than 6.0 keV show clear substrate contributions from optical excitation of the 6H-SiC substrate.

Figure 7.6 shows LEEN spectra from another FET number 22 on GS-15 from die (8,4). This particular device was tested to catastrophic failure. The LEEN spectra shown are very similar to the spectra in Figure 7.5. The main difference is that the feature associated with a quantum red-shift does not decrease in intensity as the beam voltage is increased. This behavior is difficult to explain because the expectation is for the red-shift to decrease as more current directly excites the quantum well region. However, it could be related to Si diffusion across the interface after the local heating due to failure. AES
experiments in the following section show that very high temperatures were present in the material during failure. The AlGaN NBE luminescence behaves as expected. The AlGaN NBE luminescence intensity is weak on wafer GS-15. This could be from the thickness of the silicon nitride and low current injection density or non-radiative surface traps competing for the carriers.

Figure 7.6. LEEN spectra of HFET number 22 from die (8,4) on GS-15. This testing on this FET resulted in catastrophic device failure. The AlGaN band edge luminescence contribution is clearly visible as is a red-shifted GaN band edge feature. The GaN feature associated with quantum shifting appears to be dominant near the top of the FET structure, but seems to contribute strongly at all beam energies. Spectra at energies higher than 6.0 keV from beyond the AlGaN barrier show clear substrate contributions from optical excitation of the 6H-SiC substrate.
Spectra taken from devices on GS-17 exhibit quite different characteristics from GS-15. Figure 7.7 shows LEEN spectra taken from die (6,8) transistor number 23 representative of GS-17. No device from GS-17 produced a FET with RF results. One big difference from GS-15 to GS-17 is the distinct lack of red-shifting in the GaN NBE luminescence on GS-17. The GaN NBE emission intensity is also lower, producing no signal at all at 3.0 kV beam voltages. The AlGaN NBE intensity is essentially absent in these samples. The absence of any AlGaN NBE signal could indicate non-radiative traps in the barrier layer or poor quality material. This sample also shows relatively high YL intensity at 2.20 eV for 5.0 kV beam voltages. CASINO simulations suggest that the electron beam is just beginning to penetrate the AlGaN/GaN interface at 5.0 kV.
Figure 7.7. LEEN spectra of HFET number 23 from die (6,8) on GS-17. The electrical testing on this FET showed no RF operation. The AlGaN band edge luminescence is not observed. The GaN feature associated with quantum shifting does not appear in these spectra. Spectra at higher energies than 6.0 keV still show clear substrate contributions from optical excitation of the 6H-SiC substrate.

7.3 AES Mapping of a HEMT

After performing LEEN analysis on transistor number 22 from die (8,4) on GS-15, AES mapping was performed to provide insight to possible failure mechanisms other than high field points caused by geometry effects. The AES mapping is performed by measuring the differential peak-to-peak values of the core electron counts for each element identified as a function of position as described in Section 2.4 and 2.5. Figure 7.8 is a detailed SEM image of this FET. The important features are noted. The region
of failure is evident. The resulting crater penetrates to the SiC substrate. A lot of heat was produced near the failure site. The gold from the pad metal became molten and a droplet of GaN can be seen from the intense localized heating.

Figure 7.8. Detailed SEM image of GS-15 die (8,4) FET number 22. This device was tested to failure. The most important features are noted. The crater extends down into the SiC substrate.
The actual AES maps are contained in Figure 7.9. The image contains a) a scaled down SEM image of Figure 7.8 and elemental maps of b) Ga, c) N, d) Si, e) O-bound Si, f) O, g) Au, h) C, and i) Cl. The Ga can be seen as a stripe near the top right of the square that extends diagonally downward from left to right as well as some residual GaN in the crater region. The source is from unpassivated GaN and GaN from the bulk after failure. Nitrogen can be seen between the pad metal as part of the silicon nitride passivation. Both Ga and N maps confirm the presence of the GaN drop. Elemental silicon is a weak signal and can be seen as part of the silicon nitride. However, oxygen-bound silicon can be differentiated with AES from pure silicon and can be found as the dominant silicon contributor especially in the crater region where the heat created a lot of silicon oxide. The oxygen map confirms this showing extensive oxidation forming radially around the failure site as expected. The gold map shows an absence of gold in the crater site, which does not appear to be the case in the SEM image. This is due to the high temperatures involved in the failure causing Ga to float on top of the gold and oxides to form. A thin layer of ~3 nm is all that is required to block the gold signal from AES. The carbon map proves that the crater extends into the SiC substrate. Unexpectedly, a strong chlorine signal was observed via AES. The map shows that the chlorine is localized at the failure site. Chlorine is a potential contaminant from pre-metal dips in HCl and possibly ICP etching. However, this is not usually seen in these concentrations as a result of device fabrication. The chlorine signal is unique and cannot be mistaken for any other element. It is uncertain whether Cl played any role in the failure and there are no prominent luminescence features in the LEEN spectra that predict any electrical activity from it.
Figure 7.9. AES mapping of a FET that was tested to failure. The windows are maps of the following: a) SEM image, b) Ga map, c) N map, d) Si map, e) O-bound Si map, f) O map, g) Au map, h) C map, and i) Cl map. The scale is shown above.
7.4 Discussion

It has been demonstrated that differences in LEEN spectra can be correlated with RF device performance. In this case, GS-15 and GS-17 exhibit qualitative differences from wafer to wafer that can account for GS-15 working and GS-17 not working. The higher AlGaN NBE intensities observed on GS-15 indicate a potentially higher quality material or one that does not have as many competing non-radiative defects as GS-17. The presence of the red-shifted GaN NBE emission could be an indicator of higher quality material if these quantum effects can be observed at room temperature. While the energy levels at 3.26 eV can be attributed to a donor-acceptor pair transition, this is not likely because the samples do not have any intentional or background contaminants in the chamber that would produce acceptor states in quantities sufficient to observe with room temperature LEEN spectra. These samples also have a thin AlGaN layer heavily doped with Si at 1.0x10^{19}-cm^{-3}.

Cubic phase GaN also has a bandgap of 3.25 eV, but it is not likely that there is a cubic phase present in this material. A reasonable explanation of the red-shifted peak observed in GS-15 is that the carriers are swept into the triangular quantum well creating low-level injection in the 2-DEG region. As the beam voltage increases, the direct stimulus of the electron beam at greater than 4.0 kV creates high-level injection conditions which flatten out the conduction band reducing the red-shifted emission. A plausible explanation for the persistence of the red-shifted GaN NBE emission at higher energies is that it could be enhanced by Si diffusion from the heavily doped AlGaN layer above the 2-DEG layer at temperatures where the device failed. This diffusion would produce extreme band bending and even produce a red-shift on the order of the shift
observed if even 10% of the dopants moved across the barrier during heating. Regardless, this feature was present in working FET’s and not present in failing FET’s. Yellow luminescence also appeared to be localized near the AlGaN/GaN interface as seen in experiments in Section 6.3.2.

AES was also used to identify elemental chlorine impurities in a failing device. Most likely this comes from an HCl dip used to clean the surface of the AlGaN prior to contact deposition. Unfortunately, there were no LEEN spectral features directly attributable to this finding. However, useful information about the material dynamics can be seen. An estimate of what local temperatures are involved in AlGaN/GaN device failure can be made by noting the melted gold and GaN droplets because gold melts at 1064 °C and GaN melts at > 1700 °C.\textsuperscript{105,106} These combined results at least show that Cl impurities do not have significant radiative defect contributions. Noting the results from Chapter 6 and the current results, it is reasonable to hypothesize that deep level defects found in AlGaN and GaN can be modified by device fabrication and that these results can be used to change high-frequency performance of AlGaN/GaN HEMT’s. The next set of experiments on AlGaN/GaN HEMT’s is an attempt to show this.
8.1 Experimental Preparation and Custom Processing

8.1.1 Sample Preparation

In October 2001, the HEMT layers used in this experiment were grown on sapphire by MOCVD by a commercial vendor, Company C. Sample AT06 has the following structure: \( \text{Al}_{0.30}\text{Ga}_{0.70}\text{N} \ (5\text{-nm}, \text{UID})/\text{Al}_{0.30}\text{Ga}_{0.70}\text{N} \ (22\text{-nm}, N_D = 5.0 \times 10^{18}\text{-cm}^{-3})/\text{Al}_{0.30}\text{Ga}_{0.70}\text{N} \ (3\text{-nm}, \text{UID})/\text{GaN} \ 	ext{channel/buffer (UID)}/\text{Al}_2\text{O}_3. \) Figure 8.1 shows a schematic of the layer structure. AT06 is a quarter of a 2-inch wafer that is cut into four pieces during the processing into AT06-A, AT06-B, AT06-C, and AT06-D. Figure 8.2 shows a map of AT06. The bottom right corner of the wafer is the center and the shaded regions were characterized electrically and spectrally. This material was processed with standard cleanroom device fabrication processes for AlGaN/GaN HEMT’s with the exception of the gate formation. The purpose of this experiment is to monitor individual devices for defects during fabrication and to manipulate these defects by treating the AlGaN surface in UHV to various processes. The end goal is to observe correlations in RF performance with these controlled surface treatments and correlate this to the LEEN spectra obtained during the UHV processing and after fabrication.
Figure 8.1. Layer structure for AlGaN/GaN HEMT’s on sapphire for AT06.

Figure 8.2. Wafer map of AT06 showing die labels. The shaded die were measured with LEEN between fabrication steps. The dashed line indicates where the wafer is cut during processing.
8.1.2 Custom Fabrication Procedures

In order to create devices that can be modified in UHV and analyzed by LEEN to monitor the effects of processing on the gate region, a custom fabrication scheme is required. The process is identical to normal device processing up to the ohmic contact formation at which point the processes diverge. The new process includes a new robust, clean gate mask with modifications to the gate region performed in UHV. The development of the mask will be discussed later. Figure 8.3 shows a schematic of the new process. In this process, LEEN spectra will be taken in the SEM at low temperature in the middle of the gate region 1) after ohmic contact lift-off, just before the anneal and then again 2) after the ohmic contact anneal and gate mask patterning, prior to the gate metalization. LEEN spectra were also to be taken after the gate formation after the via etch, but this portion of the experiment never materialized for reasons to be discussed below.
Figure 8.3. Schematic of custom HEMT processing designed for UHV modification of the gate region while protecting the rest of the device. The new gate mask does not outgas and is stable at high temperatures.
One of the defining aspects of this new procedure is the ability to modify the gate surface to change or induce material defects in UHV. Figure 8.4 shows a schematic of what kinds of processes can be performed. This figure shows a more detailed process between the gate lithography and Schottky contact deposition than Figure 8.3. In-situ surface cleaning can be performed with inert ion sputtering using argon or nitrogen. The sample can be annealed in vacuum to desorb contaminants or anneal out damage to the lattice as shown. These procedures are expected to have an effect on the defect luminescence that should correlate to changes in device properties. The ability to perform LEEN spectroscopy on the samples in-situ allows for characterization of the samples as received, after sputtering, and after annealing.
Figure 8.4. Schematic showing UHV gate modifications. The samples can be sputter cleaned with argon or nitrogen or annealed prior to metal deposition.
8.1.3 Development of UHV Compatible Gate Mask

To make these experiments possible, a special gate mask is required with several special properties. The requirements are 1) the masking material must not outgas in UHV, 2) the mask must be chemically stable, 3) the mask must be stable at temperatures as high as 850 °C, 4) the mask must not react with AlGaN/GaN, 5) the mask must be thick enough for proper lift-off to occur, 6) the mask must be thin enough to avoid shadowing effects during sputtering, and 7) the side-walls of the mask must be vertical or undercut.

The first choice for this mask was Si$_3$N$_4$ because it is already commonly used as a passivation layer for reasons 1 – 6 listed above. Unfortunately, several attempts to get proper side-walls for good lift-off failed. The resulting mask had bowl shaped gate openings that resulted in winged gates, which shorted out the source and drain and resulted in high field points.

The next choice was to mimic the PMGI/photoresist dual layer mask that results in an undercut promoting good lift-off. Using a dual layer dielectric stack of SiO$_2$/Si$_3$N$_4$ with a combination of anisotropic and isotropic RIE etching allowed the dual layer resist technique to be copied. Using the anisotropic, Freon-23, RIE etch, the gate mask was etched down to the channel. Once at the channel, the Freon-23 etch was switched to the isotropic, Freon-14 etch. Since the Freon-14 etch preferentially attacks the Si$_3$N$_4$, the bottom layer is etched away faster forming an undercut suitable for liftoff. Figure 8.5 shows a schematic of the formation of this mask with undercut. The optimal layer structure obtained was 100 nm of SiO$_2$ on top of 200 nm of Si$_3$N$_4$ on top of the AlGaN/GaN layer. The mask was chosen to be 300 nm thick because the metal layer
thickness can be no less than 150 nm. The mask should be at least twice as thick as the gate metal for good lift-off. If the metal thickness is less than 150 nm, the step coverage over the 100 nm mesa will no longer be continuous. Ideally, the metal should be at least twice the thickness of the step height, but a reflow technique is used on the PMGI for the mesa formation that produces a smooth mesa. However, since the gate width of the FET’s of interest is 1.2 μm, the gate mask must be much thinner than this to avoid shadowing effects during UHV processing. A 300 nm thick mask was determined to have a reasonable aspect ratio for these experiments.

Figure 8.5. Dual layer dielectric gate mask formation with gate patterning. The undercut is produced from the preferential etching of the silicon nitride.
8.2 Electrical Characterization of Ohmic Contacts to AlGaN/GaN HEMT’s

Using the TLM analysis method as described in Section 2.7, the specific contact resistance and sheet resistance was extracted for all die on the AT06 wafer. The TLM measurements were performed just after the ohmic contact anneal and prior to the gate mask deposition. Figure 8.6 and Figure 8.7 show the specific contact resistance and sheet resistance respectively along Column 5 to show any radial patterns that might be evident in the wafer. The values from Column 5 are representative of the entire wafer. The figures both show increasing values from the center of the wafer to the edge or from Row 8 – 16. The contact resistance increases slowly with an apparent outlier at die (5,14) while the sheet resistance appears to increase by a factor of 1.4 across the wafer with a sharp increase beginning at die (5,15). The average specific contact resistance for this wafer is $2.7 \times 10^{-5} \, \Omega \cdot \text{cm}^2$ and the average sheet resistance is 571 $\Omega$/sq.

Figure 8.6. Specific contact resistance of AT06 along Column 5.
Figure 8.7. Sheet resistance of AT06 along Column 5.

8.3 LEEN Characterization on Individual HEMT’s in SEM

8.3.1 CASINO Predictions for AT06

As described in Section 2.5, the wafer was analyzed by LEEN Spectroscopy in the JEOL 7800F SEM during processing. The electron beam spot size used was 150 nm rastered over a 2.0 μm x 2.0 μm square area. The spectra were taken at low temperature with liquid He at 11 K. Using the SEM, LEEN spectra were taken in the middle of the gate fingers before the ohmic contact anneal and after the gate mask patterning on many die of each piece of the wafer with electron beam energies ranging from 1.0 – 3.0 keV. The depth of the electron beam penetration varied from 20 – 87 nm, as calculated from CASINO. Monte Carlo simulations of penetration depth versus beam energy are shown in Figure 8.8 for these particular structures. At 1.0 keV, the beam penetration is confined to the top AlGaN barrier layer. At 2.0 and 3.0 keV, the beam penetrates the AlGaN/GaN...
interface. As before, CASINO does not account for carrier diffusion. Figure 8.9 shows the expected contribution to the luminescence spectra for each region, AlGaN and GaN. The spectra of different HEMT gate structures are obtained for many die as a function of beam energy. Optical emissions were collected in the 1.90 – 4.13 eV photon energy range. The yellow luminescence is integrated in the 1.90 – 2.77 eV range. The donor-acceptor luminescence is integrated from 2.77 – 3.35 eV. GaN intensities are integrated from 3.42 – 3.72 eV. AlGaN intensities are integrated from 3.72 – 4.13 eV. Allowances are made for slight shifts in the energy levels of the spectral features when integrating. The phonon replica for GaN is not included in the integrals and ignored for the other features because they are small in comparison and not easily deconvolved. These integrated values will be compared with contact resistance and sheet resistance values from die to die at different beam energies.

![Figure 8.8. CASINO simulation for AT06 electron depth dependence for LEEN spectra. The surface of the sample is to the left with the AlGaN layer and GaN bulk clearly marked.](image)

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Figure 8.9. CASINO electron energy distribution integrated by region as a function of beam voltage. The AlGaN curve shows the CASINO curves integrated from 0 – 30 nm. The GaN curve shows the curves integrated from 30 nm to the bulk.

Figure 8.10 shows an SEM image of a HEMT on AT06 identifying structural features and the point of analysis. The highly textured surface is the ohmic contact metal layer. The dark smooth surface is the SiO₂/Si₃N₄ gate mask through which LEEN spectra are taken. This mask also contains the pad metal features that create large thick contact pads for device probing. The AlGaN mesa can be seen as a transparent rectangle that encompasses the source and drain ohmic metals. The undercut achieved on the gate mask can be seen as a light border on the gate fingers situated between the source and drain metal. Using the SEM, the actual gate region of HEMT’s on AT06 are analyzed prior to ohmic contact anneal and directly after the gate mask formation as shown in Figure 8.3 with the ability to return to the exact same analysis point after each step.
8.3.2 LEEN Spectral Features and Temperature Dependence

There are several spectral features that are not observable at room temperature. On sample AT06, the donor-acceptor pair luminescence is not detectable until the sample is below 100 K. Figure 8.11 shows a temperature dependent plot of LEEN spectra at 3.0 keV for AT06 die (5,8) to demonstrate the features that are commonly observed.\textsuperscript{110} The spectra are plotted from low temperature at 10 K to room temperature from top to bottom. Once the sample is cooled down, the main features can be resolved and identified.
Centered at 3.962 eV is the AlGaN donor bound exciton with the LO phonon replica at 3.869 eV. The AlGaN donor bound exciton can change energy with Al mole fraction and strain. The GaN donor bound exciton is observed at 3.479 eV with the associated LO phonon replica at 3.388 eV. The donor-acceptor-pair (DAP) transition is found at 3.281 eV with 3 LO phonon replicas at 3.188, 3.104, and 3.010 eV. The yellow luminescence, which is normally observed at 2.20 eV, is observed here centered at 2.4 eV. The deep level defects observed with low temperature LEEN spectra will be used to characterize ohmic contacts and device performance in the next sections. A direct comparison of ohmic contact properties will be made between DAP, YL, sheet resistance, and contact resistance.

![Temperature dependent spectra from AT06 die (5,8) prior to the ohmic contact anneal. The spectra are taken from low temperature at 10 K to room temperature from top to bottom. Salient features are the YL, donor-acceptor pair luminescence with phonon replicas, GaN near band edge and AlGaN near band edge with associated phonon replicas.](image)

Figure 8.11. Temperature dependent spectra from AT06 die (5,8) prior to the ohmic contact anneal. The spectra are taken from low temperature at 10 K to room temperature from top to bottom. Salient features are the YL, donor-acceptor pair luminescence with phonon replicas, GaN near band edge and AlGaN near band edge with associated phonon replicas.
Rather than show 78 spectra from 3 beam energies at 13 die for 2 process steps, the data has been condensed into several plots showing radial trends along Column 5 for specific features in the LEEN spectra. The following convention is used for Sections 8.3.3, 8.3.4, and all of 8.4: The spectra at Point (1) in Figure 8.3 prior to the ohmic contact anneal are denoted by open shapes. The annealed spectra after the gate mask patterning and etch at Point (2) are denoted by solid shapes. Spectra at 1.0, 2.0, and 3.0 keV are denoted by black squares, red circles, and green triangles respectively. Spectral features examined include AlGaN NBE intensity, AlGaN and GaN NBE energy level, AlGaN and GaN NBE FWHM. Defect emission intensity for YL and DAP are also plotted. All intensity measurements are normalized to GaN band edge emission intensity. Finally, the DAP is compared directly to the YL instead of GaN NBE intensity. For clarity, the spectral information is plotted in groups. For all electron beam energies, the LEEN spectra taken before the ohmic contact anneal are combined on one plot in Section 8.3.3. The LEEN spectra taken after the gate mask patterning are combined on a separate plot in Section 8.3.4. The combined data is then presented to compare the effects of processing on the defect emission in Section 8.4.1.

8.3.3 LEEN Characterization Before Ohmic Contact Anneal

8.3.3.1 AlGaN NBE Radial Degradation

The LEEN spectral features discussed in this section were obtained just after the ohmic contact lift-off and just prior to the ohmic contact anneal. The spectral features
should ultimately provide some insight to the quality of the material and contacts to it. Figure 8.12 shows the AlGaN NBE intensity normalized to GaN NBE intensity for Column 5 at 1.0, 2.0, and 3.0 keV beam energies. The general trend is a slight decrease in AlGaN NBE intensity radially from the center of the wafer to the edge. The curves for all energies track each other reasonably well. The 1.0 keV curve has a factor of 2 higher intensity than the 2.0 keV curve, which in turn is a factor of 2 higher than the 3.0 keV curve. This behavior is expected from the theoretical CASINO simulations in Figure 8.9.

The AlGaN NBE energy level is plotted in Figure 8.13 for Column 5 for all electron beam energies. In this plot, all AlGaN NBE energy levels are identical within error for a given die at all electron beam energies. The AlGaN NBE energy level significantly decreases monotonically from the center of the wafer to the edge by about 0.06 eV indicating a non-negligible decrease in Al mole fraction.

The AlGaN FWHM is also plotted in the same fashion in Figure 8.14. The AlGaN NBE broadening may be an indicator of material quality or compositional changes in the AlGaN layer. In this plot, the FWHM is the same for all electron beam energies. The FWHM does not change appreciably across the wafer, but some scatter in the data is observed at higher row numbers. This may be related to the decrease in AlGaN NBE signal intensity at higher row numbers. However, all of the FWHM data is within a 0.017 eV (λ = 1.4 nm) window.
Figure 8.12. AlGaN NBE intensity normalized to GaN NBE intensity for AT06 Column 5 prior to the ohmic contact anneal. The AlGaN NBE intensity decreases from the center radially outward.

Figure 8.13. AlGaN NBE energy level for AT06 Column 5 prior to the ohmic contact anneal. The AlGaN NBE energy level decreases monotonically from center to edge signifying a reduction in Al mole fraction.
Figure 8.14. AlGaN NBE FWHM for AT06 Column 5 prior to the ohmic contact anneal. The FWHM does not change significantly and is tightly grouped.

8.3.3.2 GaN NBE Radial Consistency

The GaN NBE energy level is shown in Figure 8.15. The GaN NBE decreases only 0.001 eV (λ = 0.1 nm) across the entire wafer for 2.0 and 3.0 keV electron beams. This decrease is within the limits of the monochromator’s resolution capabilities. At 1.0 keV the GaN NBE energy level is scattered, but still only varies by 0.006 eV (λ = 0.6 nm). The slight scattering at 1.0 keV may be attributed to the AlGaN/GaN interface where Al diffusion at this interface and strain between the layers may cause lattice changes in the GaN, since any signal from GaN at 1.0 keV must be from carrier diffusion across the AlGaN and recombining at the 2-DEG layer. The changes in the GaN NBE at 1.0 keV could also be related to quantum shifting due to band bending at the interface. The GaN FWHM is shown in Figure 8.16. This plot shows no detectable variation in
FWHM at 2.0 and 3.0 keV but small scatter in the data at 1.0 keV. Again, all the data falls within a 0.017 eV ($\lambda = 1.7$ nm) window, but neglecting the 1.0 keV data, there is no measurable difference. The variance in GaN FWHM at 1.0 keV is also attributed to the AlGaN/GaN interface strain, defects, and more likely quantum shifts.

Figure 8.15. GaN NBE energy level for AT06 Column 5 prior to the ohmic contact anneal. At 2.0 and 3.0 keV, the GaN NBE energy level is constant. At 1.0 keV, there is more variation, which is still small (~0.004 eV).
Figure 8.16. GaN NBE FWHM for AT06 Column 5 prior to the ohmic contact anneal. At 2.0 and 3.0 keV, the FWHM is a constant. There are small variations at 1.0 keV.

8.3.3.3 GaN DAP and YL Defects Radial Increase and DAP/YL Consistency

The most interesting measurements are the direct observation of defects found in the bandgap of AlGaN and GaN. The luminescence from the GaN DAP has been normalized to the GaN NBE intensity and plotted as a function of position in Figure 8.17. The normalized DAP intensity increases from the center to the edge of the wafer at 1.0 keV, which is attributed to the GaN at the AlGaN/GaN interface. The normalized GaN DAP intensity for 2.0 and 3.0 keV appear to have approximately the same values, which also increase across the wafer. The GaN DAP features from the 2.0 and 3.0 keV spectra have a lower intensity than the 1.0 keV spectra, strongly implying that the source of the GaN sub-band defect luminescence is localized at the 2-DEG interface. There is about an order of magnitude decrease in the normalized GaN DAP intensities from 1.0 to 2.0 and
3.0 keV spectra. The data point for die (5,13) at 3.0 keV was discarded due to an obvious error in the measurement. The same measurement was performed for the GaN YL. The results are plotted in Figure 8.18. The GaN YL follows the same pattern as the GaN DAP except the luminescence intensity is about a factor of 4 – 5 lower. This is evident when comparing the GaN DAP directly to the GaN YL by taking the ratio of the DAP to YL. In Figure 8.19, the GaN DAP to GaN YL ratio is a constant for all electron beam energies at all die. This is not the expected behavior. Instead, some variation as a function of position was anticipated due to local variations in defect distributions that should ultimately affect contact performance.

Figure 8.17. GaN DAP normalized to GaN NBE for AT06 Column 5 prior to the ohmic contact anneal. The DAP increases from center to edge and is strongest at 1.0 keV nearest to the surface. At 2.0 and 3.0 keV, the DAP increases very slowly. The 2.0 and 3.0 keV curves are approximately the same.
Figure 8.18. GaN YL normalized to GaN NBE for AT06 Column 5 prior to the ohmic contact anneal. The YL increases from center to edge and is strongest at 1.0 keV nearest to the surface. At 2.0 and 3.0 keV, the YL increases very slowly. The 2.0 and 3.0 keV curves are approximately the same.

Figure 8.19. GaN DAP normalized to GaN YL for AT06 Column 5 prior to the ohmic contact anneal. For all energies and die, the value is identical to within error.
8.3.4 LEEN Characterization After Ohmic Contact Anneal and UHV Gate Mask

The LEEN spectral features compared in this section are taken after the ohmic contact anneal and gate mask patterning, prior to gate metal deposition. In Figure 8.20, the AlGaN NBE intensity when normalized to the GaN NBE intensity now shows an increase from the center of the wafer to the edge. In general, the AlGaN NBE intensity is lower. There is no pattern in the spectra related to electron beam energy other than the 2.0 and 3.0 keV spectra having the same values while the 1.0 keV spectra has more scatter in the values.

8.3.4.1 AlGaN NBE Radial Degradation After Processing

The AlGaN NBE energy level is plotted in Figure 8.21. The AlGaN NBE energy level decreases across the wafer from 3.96 – 3.91 eV. The values are independent of electron beam energy, but the AlGaN NBE energy level is less well defined from Rows 13 – 16. The decrease in AlGaN NBE energy level still indicates a decrease in Al mole fraction in the AlGaN layer. The AlGaN NBE FWHM follows a similar pattern. Figure 8.22 shows the AlGaN FWHM increasing from the wafer center to the edge. The FWHM for Rows 8 – 11 changes by only 0.020 eV (λ = 1.6 nm), while Rows 13 – 16 change by 0.080 eV (λ = 6.5 nm). There is no depth dependence observed in the AlGaN FWHM.
Figure 8.20. AlGaN NBE intensity normalized to GaN NBE intensity for AT06 Column 5 after the ohmic contact anneal and gate mask etch. The AlGaN NBE intensity appears to increase overall from Row 8 to Row 16.
Figure 8.21. AlGaN NBE energy level for AT06 Column 5 after the ohmic contact anneal and gate mask etch. The AlGaN NBE energy level decreases from center to edge signifying a reduction in Al mole fraction. The AlGaN NBE energy level from Rows 13 – 16 has more variation after annealing than Figure 8.13 suggesting possible reactions at the interface.

Figure 8.22. AlGaN NBE FWHM for AT06 Column 5 after the ohmic contact anneal and gate mask etch. The AlGaN NBE FWHM increases from center to edge. The AlGaN NBE FWHM from Rows 13 – 16 appears to have a larger increase again suggesting a possible reaction at the interface.
8.3.4.2 GaN NBE  Radial Consistency and Energy Level Shift

The GaN NBE energy level remains a constant as in Figure 8.23. The values for all electron beam energies are the same with the exception of a few scattered points at 1.0 keV. The GaN NBE energy level is at 3.471 eV and varies by 0.007 eV ($\lambda = 0.7$ nm) across the wafer. The GaN NBE FWHM shows the same pattern as the NBE energy level in Figure 8.24. The FWHM varies by 0.015 eV ($\lambda = 1.5$ nm) while remaining centered around 0.030 eV. The 1.0 keV curve has more variation than the higher energy curves. Variations in both the GaN NBE energy level and FWHM indicate changes in material quality. The spectra measured with 1.0 keV electron beams are especially sensitive to the GaN at the AlGaN/GaN interface.

![Figure 8.23. GaN NBE energy level for AT06 Column 5 after the ohmic contact anneal and gate mask etch. The GaN NBE is a constant with more scatter in the data for 1.0 keV electron beams.](image-url)
Figure 8.24. GaN NBE FWHM for AT06 Column 5 after the ohmic contact anneal and gate mask etch. The FWHM is constant across the wafer for all beam energies.

8.3.4.3 GaN DAP, YL, and DAP/YL Radial Increase After Processing

The LEEN spectral features associated directly with defects show patterns similar to the features prior to the ohmic contact anneal. In Figure 8.25, the GaN DAP luminescence intensity increases radially across the wafer. The spectra at 1.0 keV show the strongest increase. The increase is a factor of 25. The curves at 2.0 and 3.0 keV show smaller increases in intensity while tracing each other. The GaN YL is plotted in Figure 8.26. The YL follows the same pattern as the DAP only the increase across the wafer is now only a factor of 3. Comparing the ratio of the GaN DAP to YL luminescence intensity in Figure 8.27 shows that the DAP does indeed increase faster than the YL from Row 8 – 16. There is an apparent spike in the DAP/YL plot at Row 14. Using the defect luminescence mapping, the results can now be applied to the electrical measurements performed in Section 8.2.
Figure 8.25. GaN DAP normalized to GaN NBE for AT06 Column 5 after the ohmic contact anneal and gate mask etch. The DAP increases from center to edge and is strongest at 1.0 keV nearest to the surface. At 2.0 and 3.0 keV, the DAP increases slowly. The 2.0 and 3.0 keV curves are approximately the same.

Figure 8.26. GaN YL normalized to GaN NBE for AT06 Column 5 after the ohmic contact anneal and gate mask etch. The YL increases from center to edge and is strongest at 1.0 keV nearest to the surface. The 2.0 and 3.0 keV curves are approximately the same.
Figure 8.27. GaN DAP normalized to GaN YL for AT06 Column 5 after the ohmic contact anneal and gate mask etch. The DAP to YL ratio increases radially across the wafer with an apparent spike at Row 14. These results are vastly different than the data in Figure 8.19 before the ohmic contact anneal.

8.4 Correlation of LEEN Spectra with Processing and Electrical Measurements

8.4.1 Defect Luminescence Correlation with Fabrication and Processing

The importance of device fabrication effects on the material system cannot be neglected. Ohmic contact anneals at 850 ºC can certainly cause changes in the AlGaN/GaN interface and low temperature processing on the devices has been shown to have an effect on the material as well in Section 6.3.3. There are also changes in the surface defects of the AlGaN layer induced by the RIE step used to pattern the gate mask in these experiments. All of these effects can be observed with LEEN characterization. The data from Sections 8.3.3 and 8.3.4 have been compiled such that the effects of processing are visible on the same plot to demonstrate the extent of the changes in the
LEEN spectra due to processing alone. The processing that occurs from one set of spectra to the next is: 1) ohmic contact anneal at 850 °C for 30 seconds in nitrogen, 2) gate mask deposition, 3) gate mask lithography with a PMGI cure step at 270 °C for 5 minutes, 4) gate mask formation by RIE, and 5) solvent clean for photoresist and 1165 strip to remove PMGI. Essentially, a direct comparison of the spectra will show the effects of the ohmic contact anneal convoluted with the RIE plasma damage to the surface of the AlGaN.

8.4.1.1 Fabrication Effects on AlGaN NBE Features

Figure 8.28 shows the change in AlGaN NBE intensity during device fabrication. The intensity decreases over an order of magnitude near the center of the wafer, but decreases less near the edge of the wafer. This decrease in intensity is associated with the RIE etch because annealing tends to increase band edge luminescence by repairing lattice damage. Since the change in AlGaN intensity after processing is less near the edge, this may be a pattern created by the RIE having different reaction kinetics from edge effects. The various non-radiative surface defects created by the RIE may compete with the radiative AlGaN transitions. The effects of processing on the AlGaN NBE energy level are shown in Figure 8.29. Near the center of the wafer, there does not appear to be any change. However, approaching the edge of the wafer, the NBE energy level increases slightly. The NBE energy level is more scattered after the ohmic contact anneal. From the intensity plot in Figure 8.28, the values near the center of the wafer have the lowest intensity. The AlGaN FWHM has the same trend as the AlGaN NBE energy level as in Figure 8.30. After the anneal, the FWHM increases on from Rows 13 – 16 while
remaining unchanged from Rows 8 – 11. The AlGaN FWHM and NBE energy level are related since the changes causing the NBE energy level to shift could be from a broadening of the NBE emission. This change could indicate a change in strain or reaction at the AlGaN/GaN interface.

Figure 8.28. AlGaN NBE intensity for AT06 Column 5 before the ohmic contact anneal and after the gate mask etch. The AlGaN NBE intensity decreases substantially after processing with the largest change occurring near the wafer center.
Figure 8.29. AlGaN NBE energy level for AT06 Column 5 before the ohmic contact anneal and after the gate mask etch. The AlGaN NBE energy level becomes more scattered after processing. The NBE closer to the wafer edge appears slightly higher after processing.

Figure 8.30. AlGaN NBE FWHM for AT06 Column 5 before the ohmic contact anneal and after the gate mask etch. The AlGaN NBE FWHM becomes more scattered and much higher closer to the wafer edge after processing.
8.4.1.2 Fabrication Effects on GaN NBE Features

While the AlGaN NBE energy level remained the same or increased, the GaN NBE energy level decreased systematically by 0.006 eV across the wafer after the ohmic contact anneal and gate mask formation. Figure 8.31 shows the GaN NBE energy level during fabrication. Prior to the ohmic contact anneal, the GaN NBE energy level was a constant. These results strongly suggest enhanced band bending at the AlGaN/GaN interface from the removal of acceptor states attributed to the DAP and YL luminescence during the ohmic contact anneal. This enhanced band bending will produce an enhanced quantum effect by red-shifting the observed GaN band edge emission. In Figure 8.32 the GaN FWHM is shown to change from a constant value of 0.025 eV to a scattered plot with values from 0.025 – 0.035 eV for 2.0 and 3.0 keV electron beams. The 1.0 keV curves were scattered at all stages of the processing. The curves at all electron beam energies show that the FWHM increases slightly after the ohmic contact anneal and gate mask formation. This inhomogeneous broadening may be from a change in the strain by modifying the interface with the anneal.
Figure 8.31. GaN NBE energy level for AT06 Column 5 before the ohmic contact anneal and after the gate mask etch. The GaN NBE energy level decreases systematically after processing.

Figure 8.32. GaN NBE FWHM for AT06 Column 5 before the ohmic contact anneal and after the gate mask etch. The GaN NBE FWHM becomes scattered after processing. The FWHM increases slightly near the wafer center and does not change near the edge.
8.4.1.3 Fabrication Effects on Deep Level Defects

The largest change in spectral data is observed in the deep level defects associated with the GaN. All of the deep level defect luminescence decreased dramatically during processing. This is a result of the 850 °C anneal for 30 seconds to form the ohmic contacts. Figure 8.33 shows the change in the GaN YL. The YL drops by a factor of 10 for all electron beam energies during processing. The relative intensities between the spectra at different electron beam energies remain the same. The 1.0 keV curves always exhibit the highest GaN YL intensities and the 2.0 and 3.0 keV curves match each other for each step in the process. This indicates that the defects are localized at the AlGaN/GaN interface and the interface defect localization does not change during the fabrication, but the concentration does change. The YL decreases as expected from the anneal by a factor of 10 across the wafer. The decrease in YL is uniform. On the other hand, the GaN DAP does not decrease uniformly. Figure 8.34 shows the processing effects on the GaN DAP. While a decrease is expected from the ohmic contact anneal, there is not a uniform decrease as with the GaN YL intensity. In this case, the material started out with the DAP luminescence intensity at 1.0 keV at 4 – 5 times the 2.0 keV and 3.0 keV. However, after processing, the DAP decreased by a factor of 4 – 5 near Row 8 while not decreasing at Row 16. After processing, the 1.0 keV curves still exhibit the highest DAP while the 2.0 and 3.0 keV curves track each other.

This is a clue about the nature of these defects. The GaN YL and DAP are localized at the AlGaN/GaN interface. The YL decreases uniformly across the wafer during processing, and yet the DAP does not. The DAP decreases less in the regions where the Al mole fraction has been measured to be lower. The lower Al mole fraction
areas have a lower lattice mismatch and less strain. It is possible that this lack of mismatch makes the areas with lower Al content more stable than the areas with higher strain. One hypothesis is that the deep level defects responsible for the YL in GaN are not affected by the strain effects and the deep level defects responsible for the DAP in this GaN are affected. The resulting change in the piezoelectrically generated field could affect the movement of these defects differently during the ohmic contact anneal. The fact that the anneal reduces the observed concentration of defects also strengthens the idea that the YL and DAP luminescence are caused by patches of small volumes of concentrated point defects.

Figure 8.33. GaN YL normalized to GaN NBE intensity for AT06 Column 5 before the ohmic contact anneal and after the gate mask etch. The GaN YL significantly decreases during processing while maintaining the same pattern. The GaN YL increases from the center to the edge of the wafer.
Figure 8.34. GaN DAP normalized to GaN NBE intensity for AT06 Column 5 before the ohmic contact anneal and after the gate mask etch. The GaN DAP significantly decreases during processing near the center of the wafer while changing little near the edge. The GaN DAP increases for increasing Row number.

Comparing the changes in the GaN DAP to YL intensities, Figure 8.35 shows an increase in DAP to YL ratio during processing. This is due to the YL decreasing faster than the DAP. Prior to the ohmic contact anneal, the DAP to YL ratio was constant across the entire wafer. After the ohmic contact anneal and gate mask formation, the data is no longer constant but increases slowly from Row 8 – 16 with a spike at Row 14. The data also has more scatter after processing than the un-annealed data. The change in DAP to YL ratios indicates that a more accurate assessment of expected device parameters will be obtained by performing measurements after processing. LEEN spectra demonstrating these changes for die (05,08) on AT06 at 3.0 keV are shown in Figure 8.36. The GaN YL decreases dramatically as a function of processing while the
GaN DAP decreases marginally. Die from closer to the center of the wafer show a larger decrease in DAP as a function of processing. An interesting feature other than the decrease in YL is the shift in energy. The GaN YL shifts from being centered at 2.43 eV to 2.35 eV after the gate mask fabrication. This shift in energy is most likely a result of energetic species in the RIE physically reacting with the AlGaN surface. This theory will be corroborated by data in the UHV processing section to follow. Both of these YL energy levels are higher than the reported literature values for GaN YL at 2.20 eV. It is possible that the YL being observed in these experiments is the AlGaN equivalent of GaN YL, meaning that the deep levels responsible for the 2.20 eV luminescence in GaN are manifesting themselves as a slightly deeper level in the AlGaN as shown by Bradley et al. The fact that the YL intensity and energy level decreases as the AlGaN decreases could indicate that the higher energy deep level defect emissions originate from the AlGaN. If the YL is a convolution of GaN and AlGaN YL-related defects, then a reduction in the luminescence intensity by inducing non-radiative defects from sputtering could lower the AlGaN contribution as well, shifting the YL energy level to the intrinsic GaN value. The YL intensity having the largest relative contribution to the spectra at 1.0 keV spectra is also suggestive.
Figure 8.35. GaN DAP normalized to GaN YL for AT06 Column 5 before the ohmic contact anneal and after the gate mask etch. The GaN DAP to YL ratio changes from a constant to a slowly increasing plot with a spike at Row 14 during processing.

Figure 8.36. LEEN spectra at 3.0 keV of die (05,15) on AT06 before the ohmic contact anneal and after the gate mask etch. The LEEN spectra shows a strong decrease in GaN YL and a decrease in GaN DAP. The GaN YL shifts to lower energy after processing.
8.4.2 Defect Luminescence Correlation with Electrical Contact Measurements

The most significant findings in this data set are the direct correlation of ohmic contact properties with LEEN spectra taken at the gate region of individual devices. Figure 8.37 compares the AlGaN NBE energy level with the sheet resistance as a function of wafer position. The unprocessed AlGaN NBE energy level is an excellent indicator of the sheet resistance. As the AlGaN NBE energy level decreases monotonically across the wafer from center to edge, the sheet resistance increases in the same fashion. This pattern is most likely from radial temperature variations in the wafer during the AlGaN growth resulting in a gradient of Al mole fraction that decreases closer to the edge of the wafer. The change in sheet resistance as a function of AlGaN NBE energy level is shown in Figure 8.38 for a 3.0 keV beam. For small changes in Al mole fraction, the sheet resistance decreases linearly with an increase in AlGaN NBE energy level agreeing with the results in Section 6.3.4. The sheet resistance does not correlate with any of the normalized GaN YL or DAP measurements. This indicates that the sheet resistance does not critically depend on these defects. Rather, the most influential physical parameter on the sheet resistance is the Al mole fraction.
Figure 8.37. Sheet resistance and AlGaN NBE energy level as a function of position on AT06. The square, circle, and triangle shapes represent the AlGaN NBE energy level at 1.0, 2.0, and 3.0 keV respectively on the left side axis. The sheet resistance is represented by the cross marks and uses the right side axis. As the AlGaN NBE energy level decreases, the sheet resistance increases.

Figure 8.38. Plot of sheet resistance as a function of AlGaN band edge energy level for AT06 at 3.0 keV. The sheet resistance decreases as the AlGaN band edge energy level increases consistent with the expectation of a higher sheet charge density with higher Al mole fraction.
While the deep level defects associated with the DAP and YL luminescence in GaN do not appear to have any effect on the sheet resistance, they do correlate well with ohmic contact resistance. Figure 8.39 shows the GaN DAP to YL ratio and contact resistance as a function of position on AT06. The GaN DAP to YL ratio is included for 1.0, 2.0, and 3.0 keV electron beams after the ohmic contact anneal and gate mask formation. The ohmic contact resistance appears to track very well with the DAP to YL ratio across the wafer with the spike in the luminescence measurements at Row 14 corresponding to a jump in contact resistance at the same position. The GaN DAP to YL ratio curve at 1.0 keV has the highest correlation with the ohmic contact resistance curve. A plot of the GaN DAP to YL ratio as a function of contact resistance for the 1.0 keV case is shown in Figure 8.40. The DAP to YL ratio increases as the ohmic contact resistance increases. Since the 2.0 and 3.0 keV curves do not have the same high correlation as the 1.0 keV case, the defects closest to the ohmic contact near the surface have the largest effect on the contacts.

Since the DAP to YL ratio is constant across the wafer prior to the ohmic contact anneal, it is clear that the defects must be subjected to the full fabrication process before they can yield device specific qualitative information about the ohmic contact resistance. From the deep level defect spectra, the GaN YL and DAP both increase from the center to the edge of the wafer before and after processing. From this result alone, the contact resistance is expected to increase in the same fashion. The YL is associated with a deep level acceptor ~1.1 eV above the valence band. The DAP is associated with a shallow level (shallow compared to 1.1 eV) acceptor ~0.180 eV above the valence band. Both
acceptors are able to compensate the intrinsic n-type GaN to some degree. The change in contact resistance is related to clusters of regions containing point defects associated with these luminescence emissions that can locally pin the Fermi level at the defect energy level. In large concentrations, one type of defect will dominate the contact resistance. However, in small concentrations as in the case of good material, the YL related acceptors will reduce the effective barrier height for electrons passing through areas with other acceptors. Hence, the ratio of DAP to YL intensity is a better indicator of contact performance than comparing only the DAP or the YL directly to the ohmic contact resistance. An explanation for this relationship will be shown in Section 9.4.

Figure 8.39. Ohmic contact resistance and GaN DAP to YL ratio as a function of position on AT06. The square, circle, and triangle shapes represent the GaN DAP to YL ratio at 1.0, 2.0, and 3.0 keV respectively on the left side axis. The contact resistance is represented by the cross marks and uses the right side axis. As the GaN DAP to YL ratio increases, the contact resistance increases. The 1.0 keV curve tracks the contact resistance very well.
Figure 8.40. GaN DAP to YL ratio at 1.0 keV as a function of ohmic contact resistance for AT06. Increasing DAP to YL ratio correlates well with contact resistance.

8.5 UHV Gate Processing

After Point (2) in Figure 8.3, the gate mask has been formed and the wafer is diced into the four sections, AT06-A, AT06-B, AT06-C, and AT06-D. At this point, the samples are ready for gate metal deposition and pre-metal deposition surface treatments to determine the effectiveness and change in deep level defects that each type of modification has on the devices. The first sample was subjected to a control run consisting of only the standard 1:10 HCl:DI dip for 30 seconds followed by rinsing in DI water prior to metal deposition. The next sample was not subjected to chemical processing, only a light Ar sputter to remove the oxygen and carbon contamination that is present in all devices processed in atmospheric conditions. The final sample was subjected to the same sputter treatment followed by a rapid thermal anneal to repair the
surface damage induced by the Ar sputtering. All samples use a Pt (40-nm)/ Ti (5-nm)/
Au (105-nm) layer structure to form the gate. The Pt is used to form the Schottky barrier.
The Ti is used to promote adhesion between the Pt and Au layers. The Au is used for an
electrical contact. There is no anneal performed after the Schottky contact deposition.
The base pressure in the chamber prior to these experiments was 6.5x10^{-11} Torr.

8.5.1 HCl Pre-metal Dip Case

The control sample, AT06-C, was subjected to the standard gate metalization
scheme. Currently, this procedure is a 30 second dip in 1:10 HCl:DI followed by a DI
rinse and then metalization. AT06-C received the pre-metal dip and was placed in the
UHV chamber at 6.5x10^{-11} Torr. Room temperature LEEN spectra were taken of the
sample at 0.5, 2.0, and 3.0 keV prior to metalization. Figure 8.41 shows the LEEN
spectra of AT06-C. The salient features are the AlGaN and GaN NBE emission at 3.92
and 3.41 eV and GaN YL emission at ~ 2.25 eV. All of these spectra are normalized to
GaN NBE intensity. At the surface of the sample from the 0.5 keV electron beam, the
YL is about 6 times the GaN NBE intensity, compared to 2.0 and 3.0 keV spectra where
the YL is one tenth of the GaN NBE intensity. At 0.5 keV, the AlGaN NBE emission
also shows signs of deep level defects. The signal to noise ratio at 0.5 keV near the
AlGaN region is low. However, the signal is at least a factor of 4 – 5 above baseline.
These spectra deviate from the other as-received samples presented in the following
sections. The other as-received samples do not exhibit such deep level defects in the
AlGaN at the surface of the sample, and the GaN YL energy level of the as-received
samples is higher. Unfortunately, as-received spectra are not available for AT06-C because of processing constraints.

After taking the LEEN spectra, the Pt (40-nm)/ Ti (5-nm)/ Au (105-nm) Schottky contact was formed by electron beam evaporation of the Pt and Ti followed by thermal evaporation of the Au. The pressure in the chamber rose to a maximum of $1.4 \times 10^{-8}$ Torr with average pressures during the runs around $6 \times 10^{-9}$ Torr. AES measurements were taken of the sample at each step to confirm the presence and cleanliness of the metal sources after evaporation. Figure 8.42 shows the metal evaporation processing as monitored by AES. The differential Auger electron counts are plotted for the sample immediately after the HCl treatment and then after each successive metal layer, Pt, Ti, and Au. After the HCl dip, AES detects Ga and N as well as surface contaminants such as C and O from the native oxide or water vapor in the air. The elements showing features are noted in the spectra. From the AES measurements, it is clear that there are no gross contaminants in the metal layers introduced from the metal sources or the thermal processes involved in the metalization.
Figure 8.41. LEEN spectra of AT06-C after HCl dip prior to gate metal deposition. The spectra were taken at 0.5, 2.0, and 3.0 keV. The GaN YL is strongest at the surface and there appears to be AlGaN deep level defects at 0.5 keV associated with the HCl exposure.

Figure 8.42. AES spectra of AT06-C as a function of processing. The spectrum immediately after the HCl dip shows C, N, O, Ga, and possibly Si from the silicon nitride mask. The spectra immediately after the Pt, Ti, and Au depositions show only those metals with no C or O.
8.5.2 Sputter Clean Case

Sample AT06-B was treated with Ar sputtering at 1850 Volts for about 15 minutes at low current while rastering the ion beam over the entire sample prior to metalization. The sputtering was used to remove any surface contaminants or oxides prior to the gate metal deposition. Room temperature LEEN spectra were taken of the sample as-received and then after the sputter treatment. Figure 8.43 shows the LEEN spectra of AT06-B. The spectra were taken at 0.5, 2.0, and 3.0 keV. Open symbols denote sputtered material and the solid symbols denote as-received material. The spectra for 2.0 and 3.0 keV were normalized to the GaN NBE intensity, while the 0.5 keV spectra were normalized to the peak of the GaN YL intensity. Again, the main features are AlGaN and GaN NBE emissions and GaN YL emission. As seen in AT06-C, the GaN YL emission is largest at 0.5 keV near the surface. In this experiment, the sputtering has a dramatic effect on the luminescence properties. After sputtering, the GaN NBE intensity is essentially zero at 0.5 keV spectra indicating surface damage and defects competing for carriers that normally diffuse into the GaN. The sputtering also creates deep level luminescence in the AlGaN as seen in Sections 6.3.4 and 8.5.1. There are additional features formed above the AlGaN NBE that may be related to relaxation in the AlGaN barrier layer. This relaxation could result in some areas where there is no red-shift associated with the piezoelectric field generated at the AlGaN/GaN interface since the strain induced dipole may not be present in these areas. The GaN YL increases in intensity as well as shifts to lower energies after sputtering. Figure 8.44 shows a plot of the GaN YL peak energy as a function of electron beam energy and processing. Open symbols represent sputtered samples and solid symbols represent as-received material.
The GaN YL energy level shifts from as high as 2.320 eV as-received to as low as 2.204 eV after sputtering. The commonly reported value for GaN YL in PL and CL experiments on bulk GaN is 2.20 eV. The change in GaN YL energy level after sputtering varies from – 0.081 eV to – 0.109 eV from the 0.5 – 3.0 keV.

Figure 8.43. LEEN spectra of AT06-B as received and after Ar sputtering, prior to gate metal deposition. The spectra were taken at 0.5, 2.0, and 3.0 keV. Open shapes represent sputtered material and solid shapes represent as received samples. The GaN YL is strongest at the surface and there are AlGaN deep level defects induced by the sputtering. The YL also shifts to lower energy after sputtering as well as increases in intensity.
Figure 8.44. Plot of GaN YL peak position for AT06-B as received and after sputtering as a function of electron-beam energy. Open shapes represent sputtered material and solid shapes represent as received material. The GaN YL is consistently lower after the Ar sputtering.

During the sputtering on AT06-B, the C signal was monitored in real time. Decreases in the C signal were observed to decrease over a period of ~15 minutes. The sputtering process was ended before all of the C and O were removed so as not to destroy the active layer in the HFET. Figure 8.45 shows the AES spectra of various elemental regions before and after the sputtering. The areas shown are C, N, O, and Ga. In this plot, the open symbols denote sputtered material and the closed symbols denote as-received material. There is very clear indication of a decrease in C at ~250 eV. The N signal increases at ~370 eV. The O signal decreases at ~500 eV. The Ga signal does not appear to change from 900 – 1100 eV. This is clear indication that the C and O were reduced while the GaN signals were getting stronger. Since Ga did not increase in
proportion to the N, this may indicate preferential Ga removal by sputtering. Spectra for Al were not shown because they were not detected. The AES equipment used for this experiment was unable to resolve the low intensity Al signals at ~1400 eV. The low energy Al signals at 60 eV are typically convoluted with signals from many elements at that energy. In this case, Ga makes any analysis of Al based on features at ~60 eV impossible.

Figure 8.45. AES spectra of AT06-B as a function of processing. Open shapes represent sputtered material and solid shapes represent as received material. The regions of interest are C, N, O, Ga. Carbon and oxygen shows a significant reduction after sputtering while N increases. Ga appears to have no change. The sputtering removed C and O.

After sputter cleaning the sample and taking LEEN spectra, the Pt (40-nm)/ Ti (5-nm)/ Au (105-nm) Schottky contact was formed as in the previous section. During metalization, the pressure in the chamber rose to a maximum of 8.0x10⁻⁹ Torr with
average pressures during the runs around $6 \times 10^{-9}$ Torr. AES measurements taken of the sample after metal deposition were the same as those in Figure 8.42 showing the metal deposition for AT06-C indicating no contamination from the metal deposition.

### 8.5.3 Sputter Clean and Anneal Case

Sample AT06-A was treated with Ar sputtering at 1850 Volts for about 15 minutes at low current while rastering the ion beam over the entire sample. The sample was then annealed in vacuum prior to metalization. The RTA profile used to process AT06-A is shown in Figure 8.46. The temperature was ramped up to 750 °C over the period of ~3 minutes and held at that value for 30 seconds. The power to the heaters was then removed. The maximum temperature achieved during the anneal was 760 °C. The sputtering is used to clean the surface and the annealing is used to reduce some of the defects induced by the sputtering.
Figure 8.46. Rapid thermal anneal temperature profile for AT06-A. The value rises to 750 °C and holds this value for 30 seconds before the power is turned off. The peak temperature was 760 °C.

Room temperature LEEN spectra were taken of the sample during processing. Figure 8.47 shows the LEEN spectra of AT06-A as-received, after Ar sputtering, and after the RTA. The spectra were taken at 0.5, 2.0, and 3.0 keV. Open symbols denote sputtered material, solid symbols denote as-received material, and “x” symbols denote annealed material. The spectra for 2.0 and 3.0 keV were normalized to the GaN NBE intensity, while the 0.5 keV spectra were normalized to the peak of the GaN YL intensity. Again, the main features are AlGaN and GaN NBE emissions and GaN YL emission. As seen in AT06-B and AT06-C, the GaN YL emission is largest at 0.5 keV near the surface. In this experiment, the sputtering has the same dramatic effect on the luminescence properties as seen on AT06-B. The GaN NBE intensity is essentially zero at 0.5 keV spectra and the same deep level luminescence in AlGaN is observed as well as the extra
features above the AlGaN NBE. The GaN YL increases in intensity as well as shifts to lower energies after sputtering.

After annealing, the GaN YL intensity reduces to the as-received levels or lower. The GaN YL energy level decreases further for 0.5 and 2.0 keV spectra. At 0.5 keV, the GaN NBE intensity is also larger when compared to the GaN YL intensity after annealing, but is still lower than in as-received spectra. At 3.0 keV, the extra features formed in the AlGaN above and below the NBE are reduced by annealing. At 2.0 keV, the extra features formed in the AlGaN are not affected by the anneal. This indicates that some of the sputter damage has been annealed out, but not all of it. It also suggests that the majority of the lattice repair from the anneal happens in the GaN and not the AlGaN layer. This result is expected since GaN reacts at a lower temperature than AlGaN.

Figure 8.48 shows a plot of the GaN YL peak energy as a function of electron beam energy and processing. Open symbols represent sputtered samples, solid symbols represent as-received material, and “x” symbols represent annealed material. Like sample AT06-B, the GaN YL energy level shifts from as high as 2.299 eV as-received to as low as 2.246 eV after sputtering. The GaN YL energy level drops to as low as 2.242 eV after annealing. The largest overall shift in GaN YL energy level after all processing is –0.052 eV. The surface has the largest shift in GaN YL energy level. The GaN YL energy level at 3.0 keV actually increases after anneal, but never to the original as-received energy level.
Figure 8.47. LEEN spectra of AT06-A as received, after Ar sputtering, and after annealing, prior to gate metal deposition. The spectra were taken at 0.5, 2.0, and 3.0 keV. Open shapes represent sputtered material, solid shapes represent as received material, and the ‘x’ symbols represent annealed material. The GaN YL is strongest at the surface. The YL also shifts to lower energy after sputtering as well as increases in intensity. The anneal causes the YL to return to or below the original intensity. GaN YL energy level shifts are more evident in Figure 8.48. There are AlGaN deep level defects induced by the sputtering between the AlGaN and GaN NBE emission.
Figure 8.48. Plot of GaN YL peak position for AT06-A as received, after sputtering, and after annealing as a function of electron-beam energy. Open shapes represent sputtered material, solid shapes represent as received material, and ‘x’ shapes represent annealed material. The GaN YL energy level is consistently lower after the Ar sputtering. After annealing the GaN YL drops further except at 3.0 keV.

As with AT06-B, during the sputtering on AT06-A, the C signal was monitored in real time. Decreases in the C signal were observed to decrease over a period of ~ 15 minutes. Figure 8.49 shows the AES spectra of various elemental regions before and after the sputtering. The areas shown are C, N, O, and Ga. In this plot, the open symbols denote sputtered material, the closed symbols denote as-received material, and the “x” symbols denote annealed material. After sputtering, there is a decrease in C at ~250 eV. The N signal increases at ~370 eV. The O signal decreases at ~500 eV. The Ga signal does not appear to change from 900 – 1100 eV. As with the previous sputtering experiment, the C and O were reduced while the GaN signals increased with preferential
Ga removal. Again, Al signals were not resolved. After annealing the sample, there was no change in the spectra except with the O signal. The O intensity increased but was still below as-received levels. This indicates that the anneal did not remove any contaminants. Rather, oxygen in the form of water vapor outgassed from the heating block used to anneal the sample and redeposited on the sample. It is very difficult to desorb oxygen and carbon surface contaminants from AlGaN/GaN by annealing alone. So this behavior is expected. However, the AES analysis reveals a mistake in the processing which is probably a result of not properly outgassing the thermal mount prior to the experiment. If oxygen contamination from the outgassing had any effect on the LEEN spectra it could appear as a change in the energy level shifts in the YL at the 1.0 keV spectra. This may explain why the YL energy level did not decrease further after annealing the sample. There could be incorporation of oxygen into the first few monolayers of the AlGaN causing gallium vacancy complexes with the oxygen.

After processing AT06-A with sputter cleaning, annealing, and taking LEEN spectra, the Pt (40-nm)/ Ti (5-nm)/ Au (105-nm) Schottky contact was formed as in the previous sections. During metalization, the pressure in the chamber rose to a maximum of $8.0 \times 10^{-9}$ Torr with average pressures during the runs around $6 \times 10^{-9}$ Torr. AES measurements confirm a clean metal deposition.
Figure 8.49. AES spectra of AT06-A as a function of processing. Open shapes represent sputtered material, solid shapes represent as received material, and ‘x’ shapes represent annealed material. The regions of interest are C, N, O, Ga. Carbon and oxygen show a significant reduction after sputtering while N increases. Ga appears to have no change. The sputtering removed C and O. After the anneal, there is an increase in O, but no other changes.

8.6 Gate Lift-Off Failure

After the UHV processing of the gate region on sample AT06, the samples were ready for lift-off as shown in Figure 8.3. The current lift-off procedure involves peeling off the excess metal with tape and then cleaning the sample with solvents to remove any residue. In this case, the design rules did not allow for error because the gate metal thickness was ~150 nm while the mask thickness was ~300 nm. There were variations in the metal deposition thickness across the wafer that caused the excess metal to stick to the gate. When the lift-off was attempted, there were patches of gate metal that either did not come off or formed huge wings ~4 μm wide.
The original intent was to leave the dielectric mask in place if the lift-off worked correctly. However, once the wings formed and there were no devices that could be measured, the only chance to get working devices was to attempt a traditional lift-off by chemically etching the dielectric mask and floating the metal off of the sample. An attempt was made to lift off the gate metal by submerging AT06-C in 1:10 BOE:DI for 10 minutes. This had no effect. The chemicals could not react through the Pt layer. A second attempt to use ultrasonic agitation in conjunction with 1:10 BOE:DI for 5 minutes seemed to perform perfect lift-off. Under an optical microscope, the gates appeared well formed and the dielectric mask had been etched off. The ultrasonic procedure was repeated for AT06-B and AT06-A. However, when electrical measurements were attempted, the ohmic contacts disintegrated. The BOE attacked and completely reacted with the ohmic contact layers. A simple solution to this dilemma is to increase the thickness of the dielectric mask. Unfortunately, a thicker mask would increase shadowing effects during the UHV processing, especially sputtering.

8.7 Discussion

The original goal of this experiment was to quantify the effects of defects on individual microwave device performance and to modify these defects in UHV to better understand these effects. While this ultimate goal was not met, there are several very significant results obtained from these experiments. It has been shown that the defects responsible for the GaN YL and GaN DAP luminescence on actual devices directly correlate with ohmic contact resistance measurements and that AlGaN defects correlate with sheet resistance. The sheet resistance is controlled by the AlGaN properties and
defects rather than the GaN defects. Variations in contact resistance can be accounted for by the GaN DAP and YL luminescence. The results show that the defects responsible for the GaN DAP luminescence are more important than the GaN YL due to the ionization efficiency of these defect levels. Therefore, it takes many more deep level defects producing YL to have the same effect on contact resistance as a given number of defects producing the DAP luminescence.

The depth dependence of LEEN shows that the defects affecting the contact resistance are located near the surface of the GaN or at the AlGaN/GaN interface. The effects of the RIE on the AlGaN surface are physical in nature rather than impurity related and reduce the GaN NBE luminescence intensity levels by shortening the diffusion length of the minority carriers in the AlGaN layer. This is confirmed by sputtering AT06 in UHV with Ar, which is inert. The UHV experiments show that the GaN YL intensity is increased by sputtering, and the GaN NBE emission intensity decreases. The experiments also show that the GaN YL energy level shifts to lower energies after sputtering. Therefore, the ohmic contact anneal is responsible for the reduction in YL and DAP and the RIE is responsible for the shift in the YL energy level at the end of Section 8.4. The results also indicate that annealing the samples reduces the YL more than the DAP luminescence.

Annealing at 850 °C for the ohmic contact formation is observed to affect the AlGaN/GaN interface by red-shifting the GaN NBE at the 2-DEG interface. This is from acceptor removal near the interface as evidenced by LEEN spectra. The acceptor removal causes a deeper triangular well at the interface allowing for more quantum shifts in the observed energy levels. The AlGaN layer is affected because the AlGaN NBE
energy level changes along with the shape of the spectral feature. However, the data suggests that the lower temperature anneal at 760 °C affects the GaN and not the AlGaN. Spectra showing the effects of the anneal show GaN deep level defects reducing while sputter induced defects in the AlGaN film near the surface remain unchanged after annealing. Sputter induced defects in the AlGaN film near the AlGaN/GaN interface were observed to decrease.

These experiments contribute a more thorough understanding of the physical origin of these defects affecting the devices. From the Ar sputtering experiments, the GaN YL is increased.\textsuperscript{111,112} During the sputtering, oxygen and carbon are removed from the surface and the inert ions are breaking bonds in the material causing Ga vacancies. This is evidenced in the AES spectra by the increase in nitrogen signal with no increase in gallium signal, which reduces the effective Ga/N ratio. The idea that the GaN YL is attributed to gallium vacancies and gallium vacancy complexes is thus supported. After sputtering, annealing the samples reduces the GaN YL.\textsuperscript{113} Vacancies are more likely to recombine with interstitial impurities than remove substitutional impurities. GaN DAP luminescence is thought to be the result of shallow donors in GaN such as oxygen in a nitrogen site or silicon in a gallium site, or nitrogen vacancies recombining with acceptors such as magnesium or carbon in a gallium site. In the processed LEEN spectra after the ohmic contact anneal, the GaN DAP decreases less than the GaN YL. This is probably due to the removal of vacancies rather than the removal of substitutional acceptors.

The direct measurements of RF performance as a function of surface defect modification in UHV never came to fruition. The next experiment shows the correlation
of LEEN spectra with supplied working RF devices that have been characterized electrically. By correlating the deep level defect luminescence with the measured device properties, a reasonable conclusion can be drawn about the effects of changing the surface defects on device operation.
CHAPTER 9

DU14-OSU AlGaN/GaN HEMT’S RF CHARACTERIZATION CORRELATED WITH LEEN SPECTRA

9.1 Sample Preparation DU14-OSU

After the gate processing problems encountered in Section 8.6, another HEMT structure was provided by AFRL from another commercial vendor, Company D. The HEMT layers used in this experiment were grown on SiC by MOCVD. Sample DU14-OSU has the following structure: Al$_{24}$Ga$_{76}$N (25-nm, UID)/GaN channel/buffer (UID)/SiC. Figure 9.1 shows a schematic of the layer structure for DU14-OSU. The HEMT’s were fabricated and measured electrically at AFRL by AFRL personnel. Standard device fabrication was used without any UHV modifications to the gate as in prior experiments. The ohmic contact layers consisted of Ti (10-nm)/ Al (200-nm)/ Ni (50-nm)/ Au (20-nm) annealed at 850 °C in nitrogen. The gate contact is formed by electron-beam lithography with Ni (20-nm)/Au (280-nm) without annealing. A map of the wafer is shown in Figure 9.2. The shaded die were measured electrically and with LEEN. The purpose of this experiment is to monitor individual devices for the same deep level defects observed during the processing of other devices and to correlate these
deep-level defects with RF device performance. The differences in RF performance observed on this sample set will show that the same deep level defects that were changed during UHV processing will impact device performance.

Figure 9.1. Layer structure of AlGaN/GaN HEMT’s on SiC for DU14-OSU.

Figure 9.2. Wafer map of DU14-OSU showing die labels. The shaded die were characterized electrically and with LEEN. The bottom left corner is the center of the wafer.
9.2 Electrical Measurements

9.2.1 DC Measurements

The devices measured in this experiment had gates formed with e-beam lithography. The gate length of these devices is 0.5 \( \mu \text{m} \) with two 150 \( \mu \text{m} \) wide fingers for a total width of 300 \( \mu \text{m} \). The device performance and uniformity is excellent across the wafer. However, variations are still present. The average specific contact resistance is measured at 3.18x10\(^{-6}\) \( \Omega \text{-cm}^2 \) with a 29% standard deviation across the wafer. The average sheet resistance is 407 \( \Omega / \text{sq} \) with a 1% standard deviation across the wafer. Figure 9.3 shows a plot of the specific contact resistance as a function of radial distance from the center of the sample. In general, the contact resistance decreases from 3.2 – 1.6x10\(^{-6}\) \( \Omega \text{-cm}^2 \) toward the edge of the wafer, but there is a local maximum of 5.2x10\(^{-6}\) \( \Omega \text{-cm}^2 \) at 8.5 mm. The sheet resistance is shown in Figure 9.4. From center to the edge of the wafer, the sheet resistance decreases slowly from 411 – 398 \( \Omega / \text{sq} \).
Figure 9.3. Specific contact resistance plotted radially across DU14-OSU. The maximum occurs around 8.5 mm while generally decreasing toward the wafer edge.

Figure 9.4. Sheet resistance plotted radially across DU14-OSU. The sheet resistance is very uniform but decreases slightly radially toward the edge of the wafer.
For every HEMT measured, a family of DC curves with $I_{DS}$ as a function of $V_{DS}$ was generated for different gate voltages. From these curves, the saturation current, peak gain, and knee voltages can be measured. Figure 9.5 shows the DC I-V curves for the FET at die (12,14) on DU14-OSU. The peak $g_m$ is determined by measuring the maximum change in current for a given change in gate voltage while the drain current is saturated. The knee voltage is determined by extrapolating the saturation portion of the drain current curve and the linear portion of the drain current curve with the FET in the triode region to their intersection point. The value of $V_{DS}$ at this intersection is the knee voltage. The knee voltage determines the power efficiency of the device, since the FET can be biased at lower voltages and still maintain linearity during operation.

![Figure 9.5. DC I-V plot of a HEMT from die (12,14) on DU14-OSU. This plot shows a family of drain currents curves for a given gate voltage. The knee voltage is found by extrapolating the two linear regions of the I-V curve and the peak $g_m$ is found where the change in gate voltage produces the largest current change in saturation.](image-url)
The peak transconductance along with the device capacitance will ultimately determine the frequency performance of the device. Figure 9.6 is a plot of the peak transconductance across DU14-OSU. The $g_{m\text{-peak}}$ increases linearly across the wafer from 250 – 290 mS/mm. The change in intra-wafer transconductance is ~15%. This is a significant change. The knee voltage is extracted from the DC I-V curve with the gate voltage biased at $g_{m\text{-peak}}$. The knee voltage decreases linearly across the wafer. Figure 9.7 shows the knee voltage drop from 2.22 – 1.93 Volts from the center to the edge of the wafer.

Overall, the DC characterization shows that the trends in device parameters improve toward the edge of the wafer. This is not expected. From previous measurements on other wafers, the center of the wafer produces better quality material than the edge. Since every device parameter improves from 4 – 15 mm radially, all of these devices are better in every aspect than the devices closer to the center. Of course 15 mm is not quite far enough for true edge effects to occur, it is possible that the devices farther than that will not perform as well. Unfortunately, the devices in Column 14 did not have gates and could not be measured.
Figure 9.6. Peak $g_m$ for DU14-OSU plotted radially. The transconductance increases across the wafer.

Figure 9.7. Knee voltage at peak $g_m$ for DU14-OSU plotted radially. The knee voltage drops across the wafer.
9.2.2 RF Measurements

The high frequency performance for these FET’s was measured with an HP 8510 network analyzer. The \( h \)-parameters were extracted and recorded as a function of frequency from 1 – 26 GHz. The magnitude of the common-source current gain, \(|h_{21}|\), is plotted as a function of \( \log_{10}(\text{frequency}) \). A straight line is fitted through the data, and at the unity-current-gain, \(|h_{21}| = 0 \text{ dB}\), the unity-current-gain frequency, \( f_T \), is recorded. Figure 9.8 shows this procedure for a HEMT at die (12,14). The actual \( f_T \) at \( g_{m-peak} \) is plotted as a function of radial position in Figure 9.9. From this plot, \( f_T \) increases from 43.5 – 46.4 GHz across the wafer. This increase is about 7% across the wafer. A 7% change in frequency response across the wafer is small, but still not good for conventional manufacturing tolerance.

Figure 9.8. Plot of common-source-current gain, \(|h_{21}|\), as a function of frequency and extraction of unity-current-gain frequency, \( f_T \), for a HEMT on die (12,14) of DU14-OSU. For this device, \( f_T \) is 34.3 GHz.

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Using a rudimentary relationship between the frequency response, transconductance, and capacitance for any FET, the effective gate capacitance, \( c_g \), is extracted from the previous measurements of \( f_T \) and \( g_{\text{m-peak}} \) as in Equation 9.1.52

\[
 f_T = \frac{g_m}{c_g} \tag{9.1}
\]

For the current devices, this yields capacitance values from 36 – 40 pF/mm or 10.8 – 12 pF. Figure 9.10 shows the capacitance plotted radially across DU14-OSU. The capacitance increases away from the center of the wafer by \( \sim 10\% \). This is the only electrical parameter observed that does not improve across the wafer from center to edge. The capacitance plot is also very linear and without scatter. This is surprising considering the variations in both \( g_m \) and \( f_T \). These results also show that while the
capacitance increases across the wafer, it does not increase as fast as $g_m$. Otherwise, the frequency response would decrease across the wafer. This shows that improving the transconductance by 15% has a stronger impact on frequency performance than changing the capacitance by 10%.

![Figure 9.10. Plot of effective gate capacitance as extracted from $f_T$ and $g_{m\text{-peak}}$ as a function of distance on DU14-OSU.](image)

A reasonable estimation of the device capacitance can be made from a simple parallel plate capacitor model as in Equation 9.2

$$C = \frac{\varepsilon \varepsilon_0 A}{d} \quad (9.2)$$

where $C$ is the capacitance, $\varepsilon$ is the relative dielectric constant, $\varepsilon_0$ is the permittivity of free space ($8.854 \times 10^{-12}$ F/m), $A$ is the area of the capacitor, and $d$ is the thickness of the layer. The majority of the capacitance will come from the AlGaN barrier layer and the
capacitance associated with the 2-DEG below it. In these devices, a reasonable estimate is \( \varepsilon = 8.80 \) for Al\(_{24}\)Ga\(_{76}\)N, \( d = 25 \) nm for the barrier thickness, \( A = 300 \) \( \mu \)m gate width x (5 \( \mu \)m source-drain spacing + 2 \( \mu \)m LT), which gives \( C = 6.548 \) pF or 22 pF/mm in the units of Figure 9.10. This value is not unreasonable. Allowing for \( \pm 2 \) nm variation in the AlGaN thickness and variations in thickness associated with the 2-DEG spacing, the capacitance changes from 7.12 pF (23.7 pF/mm) to 6.06 pF (20.2 pF/mm) or \( \sim 15\% \) change in capacitance. This is approximately the same change seen in the capacitance as measured. Therefore, the change is probably due to a combination of slight deviations in the AlGaN thickness and changes in the average distance to the 2-DEG. There is evidence for a change in the 2-DEG as shown by the drop in sheet resistance across the wafer as the capacitance increases. The decrease in sheet resistance indicates a higher Al mole fraction with stronger band bending that will tend to confine the 2-DEG more and bring it closer to the AlGaN/GaN interface.\(^{96,115}\) Another consideration for the changing capacitance would be contributions from the silicon nitride passivation layer. Using \( \varepsilon = 7.5 \) for silicon nitride and \( d = 150 \) nm, the resulting capacitance is 0.93 pF or 3.1 pF/mm. Clearly, the silicon nitride contribution to the capacitance is small and could not possibly account for the radial increase in capacitance with any reasonable variation in layer thickness.

### 9.3 LEEN Characterization of HEMT’s in SEM

As in Section 8.3, the individual completed devices were analyzed in the gate region by LEEN Spectroscopy in the JEOL 7800F SEM. The electron beam spot size used was 150 nm rastered over a 2.0 \( \mu \)m x 2.0 \( \mu \)m square area. This square area
overlapped the dielectric passivation. However, no luminescence signal was observed as a result of this. The spectra were taken at low temperature with liquid He at 11 K for 3.0, 4.0, 6.0, and 8.0 keV beam energies. The depth of the electron beam penetration varied from 120 – 460 nm, as calculated from CASINO. Monte Carlo simulations of penetration depth versus beam energy are shown in Figure 9.11 for these particular structures. The electron beam does not penetrate through the silicon nitride passivation layer until 4.0 keV. Thus any signal seen from the material at 3.0 keV is from carrier diffusion. At 4.0 the electron beam is directly exciting the AlGaN barrier and the AlGaN/GaN interface. At 6.0 keV, the AlGaN barrier layer experiences the most excitation. At 8.0 keV, the electron beam penetrates well into the bulk. Figure 9.12 shows the expected contribution to the luminescence spectra for each region, AlGaN and GaN. The spectra of the gate structures were obtained for the electrically measured die as a function of beam energy. The same procedure outlined in Section 8.3 will be used to analyze the data. In this section, 3.0, 4.0, 6.0, and 8.0 keV spectra will be represented by squares, circles, triangles, and stars respectively. As before, the integrated values will be compared with contact resistance and sheet resistance values from die to die at different beam energies. Finally, the LEEN spectra will be correlated with frequency response.
Figure 9.11. CASINO simulation for DU14-OSU electron depth dependence for LEEN spectra. The surface of the sample is to the left with the AlGaN layer and GaN bulk clearly marked.

Figure 9.12. CASINO electron energy distribution integrated by region as a function of beam voltage. The AlGaN curve shows the CASINO curves integrated from 150 – 175 nm. The GaN curve shows the curves integrated from 175 nm to the bulk.
LEEN analysis of these samples reveals no detectable AlGaN NBE emission. So, for this sample set, there can be no analysis of the AlGaN NBE energy level to compare with sheet resistance. However, from the previous experiments, changes in the AlGaN NBE energy level can be inferred by the lowering sheet resistance as seen in electrical measurements. The lack of AlGaN NBE luminescence could indicate non-radiative traps competing for the carriers. Most likely, the lack of AlGaN NBE luminescence can be attributed to the relative number of carriers directly excited in the AlGaN region compared to the GaN region. Other passivated samples have shown similar low or non-existent AlGaN features as in Chapter 7. The beam energy is probably too diffuse to produce as strong AlGaN signature as seen from comparing the CASINO simulation integrals in Figure 9.12 with those in Figure 8.9, Figure 7.4, and Figure 6.6. These particular samples have the lowest expected AlGaN contribution to the luminescence of all samples included in this document. A set of representative LEEN spectra from die (12,14) on DU14-OSU is shown in Figure 9.13 for 3.0, 4.0, 6.0, and 8.0 keV. Prominent features include the GaN YL at 2.20 eV, GaN DAP at 3.25 eV, and GaN NBE at 3.45 eV. The GaN DAP appears to merge with the NBE at higher beam energies. This is probably from an increase in GaN NBE as the bulk is probed while the DAP does not increase as fast.
Figure 9.13. Representative LEEN spectra of die (12,14) on DU14-OSU at 3.0, 4.0, 6.0, and 8.0 keV. There is no obvious AlGaN NBE contribution. The standard GaN DAP and YL are observed. In this sample the GaN YL occurs at 2.20 eV. The GaN DAP emission are convoluted with the GaN NBE emissions at energies higher than 3.0 keV due to the rapid increase in GaN NBE intensity dominating the spectra. The DAP and YL intensities are consistently low.

The GaN NBE intensity as a function of radial position is shown in Figure 9.14 for all beam energies. Samples in the other sections show a moderate decrease radially for the GaN NBE intensity, but on DU14-OSU, there is a dramatic reduction in NBE intensity across the wafer. More specifically, the last column, Column 13, has a factor of 40 reduction in NBE intensity while the intensity on other die decreases slowly. Additionally, the curve at 3.0 keV has the lowest intensity, while the curves for 4.0, 6.0, and 8.0 keV agree well. CASINO indicates that for 3.0 keV, any electronic excitation of the GaN is from diffusion of the excited carriers. Beam energies greater than 4.0 keV can directly excite the GaN yielding stronger emission. It is possible that variation in the
silicon nitride passivation layer is responsible for this phenomenon. However, the nominal thickness is 150-nm and no more than 10% variation in the passivation layer thickness is expected. So, adding 15 – 20 nm to the thickness near the edge of the wafer might put the 4.0 keV spectra in the diffusion-only excitation regime, but 6.0 keV should easily directly excite the GaN. Another possible explanation is non-radiative defects competing for the carriers. However, since the device performance increases across the wafer, this is not an acceptable explanation. The only other explanation is a possible geometry effect reducing the light collection efficiency at that position on the wafer, which will not make a difference in the comparisons of spectral features that are normalized to each other or the GaN NBE intensity except in the case of normalizing to GaN NBE intensities that are very low.

Figure 9.14. GaN NBE intensity as a function of radial position on DU14-OSU. The GaN NBE intensity decreases from the center radially outward for all energies. The signal decreases rapidly after 12 mm.
LEEN spectra of the GaN NBE energy level can give information about doping levels, strain, or quantum shifts at the AlGaN/GaN interface. Figure 9.15 shows a plot of the GaN NBE energy level as a function of radial distance from the center for 3.0, 4.0, 6.0, and 8.0 keV electron beams. The data points for 3.0 and 4.0 keV after 12.0 mm were eliminated due to unacceptable signal-to-noise ratios for the GaN NBE intensities. The 4.0, 6.0, and 8.0 keV spectra all show the same energy level across the wafer of 3.453 ± .002 eV while 3.0 keV spectra show an average of 3.442 ± .003 eV. The shift to lower energy at 3.0 keV is .011 eV. Since the shift is only seen at 3.0 keV, it is possible that there is a donor bound excitonic feature associated with the AlGaN/GaN interface or this could be an indication of tensile strain at the AlGaN/GaN interface. Most likely this is because under diffusion-only excitation, the quantum well at the 2-DEG is unaltered giving red-shifted Franz-Keldysh recombination. When the electron beam directly excites the AlGaN/GaN region, a high level injection condition exists which can have a band-flattening effect reducing the effective Fermi level during excitation. This high level of excitation can reduce the apparent quantum shift. The relative uniformity of the GaN NBE as a function of position at 3.0 keV compared to other samples shows that this material has a very uniform and high quality AlGaN/GaN interface.
Figure 9.15. GaN NBE energy level as a function of radial position on DU14-OSU. At 4.0, 6.0, and 8.0 keV, the GaN NBE energy level is constant. At 3.0 keV, there is more variation, which is still small, as well as a red-shift due to band bending at the AlGaN/GaN interface.

Spectra of the deep level defects in DU14-OSU yield interesting results. These defect intensities will be compared to device measurements in later sections. The GaN DAP intensity normalized to GaN NBE intensity is again plotted for all electron beam energies as a function of radial distance Figure 9.16. The 3.0 keV curve shows that the GaN DAP is a factor of 10 higher than all other energies. The 4.0, 6.0, and 8.0 keV DAP curves again track each other. The die farther than 12 mm from the center of the wafer behave strangely, in that the 6.0 and 8.0 keV spectra appear to be unnaturally high. The 3.0 and 4.0 keV data points after 12 mm had to be omitted because the GaN NBE intensity was not observable. These results indicate that the defects associated with the DAP luminescence are localized at the AlGaN/GaN interface. Since the GaN NBE
Intensity data points after 12 mm behave strangely, the data normalized to the NBE intensity after 12 mm is difficult to interpret. Thus it is hard to describe any trends in the GaN DAP luminescence as normalized to the NBE intensity. The data could just as easily be projected to increase across the wafer, or decrease after reaching a local maximum.

Spectra of the GaN YL as normalized to the GaN NBE intensity are shown in Figure 9.17. The data indicates that the defects responsible for the YL are localized at the AlGaN/GaN interface. The 3.0 keV curve is again a factor of 10 higher than the YL in the 4.0, 6.0, and 8.0 keV spectra. As with the DAP data, the data for the YL as normalized to the GaN NBE is not reliable after 12 mm. The 3.0 and 4.0 keV data points beyond 12 mm have been omitted. The general trend is difficult to predict. The last three data points at 3.0, and 4.0 keV could increase or decrease.

Figure 9.16. GaN DAP normalized to GaN NBE as a function of radial position on DU14-OSU. The DAP increases from center to edge. The DAP is strongest at 3.0 keV nearest to the surface. At 3.0 keV, there appears to be a local maximum at 8.5 mm. The 4.0, 6.0, and 8.0 keV curves are very similar in magnitude and behavior.
Figure 9.17. GaN YL normalized to GaN NBE as a function of radial position on DU14-OSU. The YL increases from center to edge. The YL is strongest at 3.0 keV nearest to the surface. The 4.0, 6.0, and 8.0 keV curves are very similar in magnitude and behavior.

As with AT06, observing only the GaN DAP or GaN YL on DU14-OSU alone is not enough to characterize the material. Since this material has low amounts of either type of defect, comparing the GaN DAP to YL ratios as a function of position in Figure 9.18 shows a completely different behavior. There does not appear to be any distinctive trend for all of the curves. The 4.0 keV spectra indicate a value higher than the other curves. Looking at the previous figures, this is due to a decrease in the YL intensity at this energy. The 3.0 keV curve appears to reach a local maximum at 8.5 mm while decreasing toward the edge. These values will be compared against the contact resistance values and the frequency response for the HEMT’s on DU14-OSU.
Figure 9.18. GaN DAP normalized to GaN YL as a function of radial position on DU14-OSU. The 3.0 keV curve appears to peak at 8.5 mm while decreasing toward the edge. There does not appear to be any strong trends.

9.4 Correlation of LEEN Spectra with DC Electrical Contact Measurements

In this section, the luminescence measurements are compared directly with the DC device parameters measured in Section 9.2. As in Section 8.4.2, the GaN DAP to YL ratio and ohmic contact resistance is plotted as a function of radial position on the same graph in Figure 9.19 for all beam energies. There does not appear to be any correlation for energies greater than 3.0 keV. The 3.0 keV curve corresponding to the AlGaN/GaN interface follows the ohmic contact resistance. To better demonstrate this, the GaN DAP/YL ratio for 3.0 keV is plotted alone with the contact resistance in Figure 9.20. The contact resistance tracks well with the DAP/YL ratio. The GaN DAP/YL ratio for 3.0 keV is plotted as a function of contact resistance in Figure 9.21 to show the correlation. The GaN DAP/YL ratio increases with increasing ohmic contact resistance.
Figure 9.19. GaN DAP/YL ratio plotted for 3.0, 4.0, 6.0, and 8.0 keV with contact resistance as a function of radial position for DU14-OSU.

Figure 9.20. GaN DAP/YL Ratio plotted for 3.0 keV only and contact resistance as a function of radial position for DU14-OSU for closer comparison.
Figure 9.21. GaN DAP/YL ratio plotted for 3.0 keV only as a function of contact resistance for DU14-OSU for closer comparison. The DAP/YL increases with contact resistance.

Again, the GaN DAP to YL intensity ratios correspond to the contact resistance. The fact that the contact resistance correlates with defect luminescence from the GaN at the AlGaN/GaN interface suggests that the mechanism for forming the ohmic contacts to these HEMT’s is tunneling through the AlGaN barrier layer and conduction band bending at the AlGaN/GaN interface in addition to any tunneling through the metal-semiconductor interface. Figure 9.22 shows a schematic of ohmic contact current flow mechanisms in an n-type contact to an n-type AlGaN/GaN HEMT. Local variations in the doping just under the AlGaN/GaN interface will exponentially reduce the ability of free electrons in the metal contact to tunnel into the channel.\textsuperscript{116,117}
Figure 9.22. Schematic for ohmic contact formation to AlGaN/GaN HEMT’s. The electron must tunnel through the TiN layer formed at the metal-AlGaN interface and the AlGaN layer itself. The probability of tunneling is reduced when states at or just below the AlGaN/GaN interface create local p-type regions.

A very simple explanation for the variation in the DAP to YL intensity is the relative contribution of compensating carriers from each acceptor state. This theory assumes a uniform distribution of point defects. In n-type material, all of the acceptors associated with the DAP and YL should be ionized because the Fermi level is far above the acceptor levels in GaN. So, in an unintentionally doped GaN sample which will probably have ~5x10^{16} cm^{-3} donors, the number of electrons available for conduction assuming complete ionization of the donor is equal to the number of donors minus the total number of acceptors. This would require on the order of ~10^{16} cm^{-3} acceptors to change the conduction band position appreciably. This is not unreasonable. However, if this were the case, then the contact resistance would be related to the sum of the YL and DAP luminescence and not the ratio.
A better hypothetical explanation for the DAP/YL ratio correlation to the contact resistance revolves around the acceptor defects grouping around threading dislocations. These threading dislocations have been well documented in GaN with typical densities for these defects at $10^8 \text{ cm}^{-2}$. Woodall et al. have shown that charged defects in GaAs at misfit dislocations have formed depletion regions around the dislocations due to band bending. In GaN, the threading dislocations may promote this same behavior by attracting impurities that act as acceptors and causing impurity, dopant, and vacancy segregation. There have also been recent studies showing spatially localized luminescence correlations with threading dislocations. Further evidence for charged defects at threading dislocations is noted by UV-assisted photochemical etching of GaN. In this type of etching, the UV source is thought to excite free carriers that act as an etch catalyst. After etching, whisker structures are reportedly left behind standing on the substrate. It is thought that these whiskers correspond to the threading dislocations found in GaN. It is also known that defects and impurities can be drawn to dislocations forming Cottrell clouds. Localized PL measurements at these threading dislocations show that there is less or no band edge luminescence observed at these defects suggesting high recombination rates at the deep level defects.

If the background acceptor contamination is on the order of $10^{16} \text{ cm}^{-3}$, drawing impurities and point defects locally into the threading dislocation region could increase the concentrations to relatively high levels. It is at these defects that small volumes of the GaN could be locally p-type. These p-type regions will raise the conduction band with respect to the Fermi level locally and create an artificially high barrier for electrons to
tunnel into. Several of these regions in parallel will contribute to a higher effective contact resistance. The effects of local changes in the conduction band on contact resistance have been modeled by Tung with a circular patch model. In the limit of a very high concentration of one kind of these defects, the Fermi level could pin at the energy level for that defect. However, in high quality material, these defects are smaller in number and both can influence the position of the Fermi level. For these defects, the effective resistance for the areas more influenced by the YL defects will have a lower resistance than the areas more influenced by the acceptor states, since the effective local barrier can fluctuate between the conduction band of the p-type regions and the conduction band of the n-type regions. However, areas with no acceptor-like defects will have the lowest contact resistance. Figure 9.23 shows the patch model as it would apply to the AlGaN/GaN system. The barrier height shown in this figure is an effective scattering barrier that affects the lateral contact resistance after the electrons have tunneled into the semiconductor. The barrier would be affected by each type of acceptor. This shift in the conduction band could occur near the AlGaN/GaN interface as shown in Figure 9.22 in planes parallel to the interface.
Figure 9.23. Patch model as applied to the AlGaN/GaN system. The defects around the threading dislocations will have local p-type regions according to this model that will raise the effective contact resistance by changing the scattering barrier height, $\Phi_{B\text{-eff}}$. 

To explain the behavior between the DAP/YL ratios and contact resistance, the effective barrier height expressed as the difference between the conduction band and Fermi level can be expressed as a function of donor and acceptor concentrations for each region. Consider the conditions of charge neutrality locally in either region such that the Fermi level is not pinned to the deep levels associated with either the donor-acceptor pair associated with the DAP luminescence, or donor to deep level associated with the YL luminescence. Equation 9.3 gives charge neutrality in general by

$$n_o + N_{A}^{-} = p_o + N_{D}^{+}$$

(9.3)

where $n_o$ is the number of free conduction band electrons, $N_{A}^{-}$ is the ionized acceptor concentration, $p_o$ is the number of free valence band holes, and $N_{D}^{+}$ is the ionized donor concentration. If the regions containing either one type of acceptor or another are locally p-type, then for that region charge neutrality can be simplified to Equation 9.4.
This expression can be rewritten in terms of the density of states and Fermi-Dirac statistics by Equation 9.5

\[
N_v e^{\frac{E_F - E_V}{k_B T}} = \frac{N_A}{1 + 2e^{\frac{E_F - E_V}{k_B T}}}
\]  \hspace{1cm} (9.5)

where \( E_F \) is the energy of the Fermi level, \( E_V \) is the valence band energy, \( E_A \) is the acceptor level energy, \( k_B \) is Boltzmann’s constant, \( T \) is the temperature, \( N_v \) is the density of states in the valence band, and \( N_A \) is the acceptor concentration.\(^{17,80} \) Since this is a p-type wide bandgap material, this expression can be simplified and rewritten in terms of the Fermi level energy with respect to the conduction band energy by Equation 9.6 in the case of complete ionization for non-degenerate p-type doping, where \( \Phi_{B-p} \) is the effective barrier height for p-type material and \( E_C \) is the conduction band energy.\(^{139} \)

\[
\Phi_{B-p} = E_C - E_F = E_C + k_B T \ln \left( \frac{N_A}{N_v} \right)
\]  \hspace{1cm} (9.6)

The charge neutrality given in Equation 9.3 simplifies to Equation 9.7 for n-type material assuming complete donor ionization with acceptor compensation.

\[
n_o = N_D - N_A
\]  \hspace{1cm} (9.7)

The same procedure from Equation 9.5 to 9.6 can be used to express the conduction band with respect to the Fermi level for an n-type semiconductor as in (9.8), where \( \Phi_{B-n} \) is the effective barrier height for n-type material and \( N_C \) is the density of states in the conduction band.
\[ \Phi_{B-n} = E_C - E_F = -k_B T \ln \left( \frac{N_D - N_A}{N_C} \right) \]  

(9.8)

Using the patch model as shown in Figure 9.23, the effective scattering barrier height can be described by Equation 9.9 assuming the p-type regions contain the impurity defects and are surrounded by n-type GaN with gallium vacancies that cause the YL.

\[ \Phi_{B-eff} = E_G + k_B T \ln \left( \frac{N_{A-DAP}}{N_y} \right) . \left( \frac{N_D - N_{A-YL}}{N_C} \right) \]  

(9.9)

Figure 9.24 shows the expected effective barrier height around the patches near threading dislocations for various compensating acceptors in the n-type GaN from gallium vacancies as a function of increasing acceptor impurity concentration in the p-type GaN regions. Increasing the YL-related acceptor concentration decreases the effective barrier height, while increasing the DAP-related acceptor concentration increases the effective barrier height. Hence, the ratio of DAP/YL luminescence is expected to increase with increasing contact resistance.
Figure 9.24. Effective barrier height for as a function of impurity concentration in p-type GaN patches around threading dislocations for various compensating gallium vacancy acceptor concentrations.

Similarly, if both regions were considered p-type and the YL and DAP related acceptors were still separated, the effective barrier height can be shown as Equation (9.10).

\[
\Phi_{B-eff} = \frac{1}{2} (E_{A-YL} - E_{A-DAP})\frac{1}{2} k_B T \ln \left( \frac{N_{A-DAP}}{N_{A-YL}} \right)
\]

(9.10)

For p-type material with the Fermi level above or below either acceptor level, Equation 9.10 can be shown to have a similar form with the change in Fermi level related to a natural log of the ratios of the acceptor concentrations. However, the regions containing YL becoming p-type material is not as likely.

There are compelling reasons to make the assumption of locally grouped patches of different defects. The first reason supporting this is that YL is not observed in spectra.
of strongly p-type GaN doped with Mg.\textsuperscript{74,82,83,140} The second reason is that it is not energetically favorable for the defects associated with the YL to form in p-type GaN.\textsuperscript{141} Therefore, in the regions that have become effectively p-type due to the migration of point defects, it is not likely that the point defects associated with YL will incorporate themselves into the more p-type volumes. Thus there will be a separation between volumes of both types of acceptors.\textsuperscript{124,125,142} Figure 9.25 show a schematic of the proposed impurity segregation model. The tunneling associated with changes in contact resistance is exponentially dependent on changes in local GaN conduction band levels, which are directly related to the Fermi level. Thus, a relationship is expected between contact resistance and ratios of the densities of each acceptor. The luminescence intensities are proportional to the densities of these localized acceptors. This is why the ratio of DAP/YL intensities shows a correlation with the ohmic contact resistance. In physical terms, this means that the contact resistance improves based on which acceptor causes the effective barrier height to be the lowest. This does not mean that YL defects reduce contact resistance. The defects associated with DAP and YL always increase contact resistance. However, when they compete in low quantities, the YL related acceptors tend to raise the conduction band in the n-type GaN regions such that the electrons have a higher probability of tunneling through the p-type regions while the DAP only decreases tunneling probability. Figure 9.26 shows the effects of increasing YL-related acceptors on the scattering barrier. In the limit of large concentrations of YL or DAP, this will not matter as seen in Section 6.3.4 where the YL was large compared to later samples. The contact resistance increases based on the dominant deep level defect in this case.
Figure 9.25. Proposed model for segregation of impurities to threading dislocations while the bulk of the material is lightly n-type containing compensating gallium vacancies.

Figure 9.26. Change in scattering barrier in the presence of high concentrations of gallium vacancies. The effective lateral scattering barrier would lower while the tunneling barrier (\(E_C-E_F\)) into the semiconductor increases.
9.5 Correlation of LEEN Spectra with Device Measurements DC and RF

The most important result of this research is the ability to correlate the effects of deep level defects on microwave device performance. These particular devices with 0.5 \( \mu \text{m} \) gate lengths operate with an average \( f_T \) of 45 GHz at \( g_{m\text{-peak}} \). In section 9.2.2, the frequency response was shown to vary by 7% radially across the wafer. In this section, the frequency response is compared to the GaN DAP to YL ratios at 3.0 keV. Figure 9.27 shows the unity-current-gain frequency as a function of GaN DAP to YL ratio. The frequency response decreases as the DAP to YL ratio increases. This indicates that the frequency response is sensitive to the deep level defects located near the AlGaN/GaN interface. This is expected from previous results showing the direct correlation of the contact resistance with the same deep level defects. Equation 9.1 indicates that increasing the effective transconductance will increase frequency response for a given capacitance. Since the geometry and intrinsic material properties seem to be limiting the capacitance, effort should be placed on reducing the contact resistance and sheet resistance to get the maximum possible transconductance.
Figure 9.27. Unity-current-gain frequency as a function of GaN DAP/YL ratio at 3.0 keV. The frequency response decreases with increasing GaN DAP/YL ratio.

9.6 Discussion

The experiments performed in this section have shown without question that the deep level defects found in GaN near the AlGaN/GaN interface in HEMT’s have effects on device performance. Unfortunately, no direct contribution to the luminescence was observed from the AlGaN barrier layer. However, it appears from the data that the most important aspect of the AlGaN/GaN HEMT’s is the AlGaN/GaN interface itself. The deep level defects in the GaN have a strong impact on the ohmic contact values by introducing scattering at the p-type barriers, which in turn affects almost every other aspect of the device. Lowering the overall contact resistance will increase gain, and thereby frequency response. The higher transconductance associated with decreases in contact resistance also reduces the knee voltages which affects power efficiency by
allowing lower bias voltages to achieve saturation current. The effects of lowering the contact resistance by changing the concentrations of deep level defects have been shown to have a greater affect on frequency response than changes in capacitance associated with shifts in the 2-DEG position from quantum well potential changes. Reducing the defects in the quantum well area, providing a deeper well, and bringing the 2-DEG closer to the AlGaN/GaN interface will only increase the capacitance to the maximum allowed by the barrier thickness geometry.
A brief summary of the major results can be given by chapters. In Chapter 5, increasing intensities of dominant GaN DAP features in LEEN spectra of AlGaN/GaN HBT’s are correlated with the increase in ohmic contact resistance. In Chapter 6, LEEN spectra of AlGaN/GaN HEMT’s from AFRL A400 samples show the dominant GaN YL intensities correlating with contact resistance. In this chapter, the fabrication effects cause a decrease in YL during annealing in nitrogen and an increase in deep level acceptors at low temperature anneals. In Chapter 7, LEEN spectral features show pronounced differences between AlGaN/GaN HEMT’s on SiC on wafers that produced working and failing RF devices. In Chapter 8, low concentrations of DAP and YL are observed in AlGaN/GaN HEMT’s on AT06. The DAP/YL ratios from spectra at the AlGaN/GaN interface increase with increasing contact resistance. The effects of annealing processes during fabrication are shown to reduce point defects. Sputtering with Ar is shown to increase GaN YL defects and induce AlGaN deep level defects. Finally in Chapter 9, low concentrations of DAP and YL deep level defects are observed to increase the contact resistance. Again, the GaN DAP/YL intensity ratio correlates well with contact resistance. Most importantly, the GaN DAP/YL deep level defects are
correlated with frequency response. Increasing DAP/YL ratios corresponded to decreases in performance.

There are several conclusions that can be drawn from this work. The optical characterization of AlGaN/GaN device layers can be used to predict contact and device performance as shown with every sample set. The localized LEEN spectra of individual transistors can be affiliated with individual device performance as opposed to general characterization of large surface areas. Localized defects can produce changes from one device to another device less than 100 $\mu$m away that manifest themselves as changes in electrical performance. The defect localization can be lateral as well as confined to specific layers. There are optical trends in various types of AlGaN/GaN from both MBE and MOCVD that hold true for both types of growth methods when related to contact performance. Namely, the deep level defects found in the GaN nearest to the AlGaN/GaN interface directly impact ohmic contact performance. Low temperature and high temperature device fabrication has been shown to affect deep level defects. AlGaN near-band edge energy level seemed to be the dominant parameter in sheet resistance while deep level defects in the GaN at the AlGaN/GaN layer did not appear to affect it. The reverse was true for the ohmic contact resistance.

The nature of the n-type ohmic contact to n-AlGaN/GaN devices appears to be tunneling which is affected exponentially by the change in barrier height from conduction band shifts in the GaN as the electron attempts to pass from the metal contact to the 2-DEG. The deep level defects measured at the AlGaN/GaN interface from the GaN cause this barrier to shift and cause potential scattering due to the p-type regions in the GaN. The yellow luminescence associated with a shallow level donor to a deep level acceptor
can, in large quantities, pin the Fermi level. The same is true for the donor-acceptor-pair transition associated with a transition from the shallow donor to the acceptor at ~180 meV above the valence band. Together, these defects form small patches that change the Fermi level to lower values in the GaN than the surrounding unintentionally doped n-type material. The acceptor at ~180 meV above the valence band will ultimately produce regions with the highest local resistance and the deep level defects producing the yellow luminescence will not be present in large enough quantities to produce a similar effect due to unfavorable formation conditions in p-type GaN. Of course, there are plenty of areas where good ohmic contacts can form without being affected by these defects. These clusters of patches of high concentrations of defects simply act as larger parallel resistors that reduce the effective current flow in the contact. This is not Schottky barrier formation and pinning at the metal-semiconductor interface. Rather this is solely an effect of potential scattering on the GaN side of the AlGaN/GaN interface in the semiconductor itself from the defect doping.

The effects of device fabrication and surface processing in UHV on the deep level defects related to the GaN YL and DAP luminescence have been observed. Sputtering and plasma etching the surface of AlGaN/GaN samples show that the intensity of the YL increases afterward supporting the idea that the YL is related to Ga vacancy complexes. Annealing the samples has been shown to reduce all radiative defects while reducing the YL-related defects more than the DAP-related defects. Lower temperature annealing at 760 °C was shown to affect the GaN side of the AlGaN/GaN interface and not the AlGaN side of the interface. The ohmic contact anneals at 850 °C were observed to affect both the AlGaN and GaN layers. This shows that repairing sputter damage or physical etch
damage to AlGaN layers requires anneals at 850 °C or greater. However, anneals at 750 °C or slightly lower may be sufficient to remove some point defects in the GaN near the AlGaN/GaN interface.

The frequency response of AlGaN/GaN HEMT’s being directly dependent on the overall capacitance and resistance makes the importance of reducing resistance as much as possible a top priority. The majority of the capacitance in these devices can be attributed to the AlGaN layer itself and the underlying 2-DEG layer. The sheet resistance and contact resistance reduction are directly related to the transconductance. In the devices measured for Al mole fractions of around 10-30%, the sheet resistance is not typically limited by radiative defects in the AlGaN. It has been shown that physical sputtering can induce defects in the AlGaN layer, but sputtering is not normally done to the actual devices. To lower sheet resistance in these devices, the Al mole fraction could be slightly increased. Increasing the Al mole fraction will only work until the defects associated with the lattice mismatched material become great enough to adversely affect the 2-DEG and possibly even the GaN deep level defects at the AlGaN/GaN interface. Lowering contact resistance can be achieved by reducing or eliminating the defects associated with DAP and YL emissions. At the very least, knowing that these defects directly affect the contact resistance should emphasize the importance of producing material with uniform densities of these defects that maintain their composition or change uniformly when put through the entire device fabrication process.

The fact that the DAP luminescence may be related to a Mg impurity should also demonstrate the importance of using reactors that have never used Mg to produce n-type AlGaN/GaN devices. Due to the proprietary nature of some of these structures,
information about the growth and reactors are not available. Since most producers of GaN have attempted to produce p-type GaN at some point, it is very probable that there is some background Mg in the reactors.

There are many factors that have not been considered or studied in this work. The nature of non-radiative defects has clearly not been addressed. There has been no work to detect atomic impurities with SIMS at device locations. The modifications to the gate region in ultra-high vacuum have not been implemented in functional devices yet. Issues of deep level defects in the AlGaN layer itself need to be studied closely. The effects of defects in AlGaN or GaN on gate leakage current and Schottky barrier height at the metal-AlGaN interface have not been studied. The following list of possible experiments to expand upon these results should not be taken as all-inclusive, but as a guide for future studies in the AlGaN/GaN material system as applied to HEMT’s.

SIMS spectra should be taken as a function of position at individual transistor locations on wafers that have been fully processed and characterized optically and electrically. SIMS spectra should also be taken as a function of processing at specific device locations. This newly acquired instrumentation will be able to detect impurity level changes in Mg, C, O, H, Cl, F, and other elements, which may be a complete surprise, as a function of position and processing. The effects of various stages in the device fabrication can be monitored on sacrificial pieces of the wafer to give detailed information about the incorporation or diffusion of impurities throughout the material. There could be interfacial contaminants, intermixing, metal diffusion from the ohmic contacts that affects the 2-DEG layer, or any number of things that cannot be detected any other way. With this technique, the physical and chemical mechanisms of the defects
as observed spectroscopically and electrically can be identified. These results can then be combined with the RF device measurements to produce a very detailed understanding of the defects that affect the AlGaN/GaN system that to this date, no group has achieved.

Another attempt should be made to reproduce the gate mask lithography for ultra-high vacuum surface science experiments so that the effect of the metal-semiconductor interface states as induced or removed by vacuum can be compared to electrical measurements. The mask can be made thicker so lift-off will not fail. Variations of annealing, sputtering, and deposition sequences should be attempted and compared with Auger and spectroscopic data.

An attempt should be made to include the effects of non-radiative defects on the electrical performance of the device. While the data shown for radiative defects is strongly suggestive, there is still scatter in the data. This scatter is likely due to the non-radiative defect contribution to the electrical measurements. The new EBIC attachment for the SEM should be helpful in imaging some of these defects. An extremely helpful tool would be to include a new test pattern on the mask that could be used to make C-V Profiling and DLTS measurements by forming mesa regions for analysis near the transistors so that non-radiative defects can be identified and quantified. Other groups have attempted to model the density of interface states in AlGaN/GaN HEMT’s by performing frequency dependent capacitance and conductance measurements. One of the weaknesses of the LEEN technique is the inability to determine defect densities.

These measurements and techniques should be applied to structures with varying Al concentrations, substrates, and growth methods. Differences observed from MBE to
MOCVD growth could provide insight to producing better devices. There are also several types of GaN substrates available such as Si, SiC, and Al₂O₃. Results from different combinations of substrates produce vastly different electrical results. It is necessary to understand this as well. The experiments in this work have already demonstrated the importance and effects of deep level defects on device performance at DC and RF. Overall, these new experiments will further improve upon the understanding of defects in AlGaN/GaN and their relation to electrical performance. The result will be the realization of high performance, reliable, and repeatable AlGaN/GaN devices that can be applied to commercial products.
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