STUDIES ON THE RELIABILITY OF NI-GATE ALUMINUM GALLIUM NITRIDE / GALLIUM NITRIDE HIGH ELECTRON MOBILITY TRANSISTORS USING CHEMICAL DEPROCESSING

By

PATRICK GUZEK WHITING

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For my family and friends
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I quite literally wouldn't be where I am if it weren't for my parents. I don't mean this in a superficial way, that they were the ones who brought me into this world, but that if it weren't for their constant support I would have never been able to achieve this. I learned to read and write because they were there to struggle with me every step of the way. So, it's no surprise that I love them.

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TABLE OF CONTENTS

ACKNOWLEDGMENTS ................................................................................................................. 4
LIST OF FIGURES .......................................................................................................................... 8
LIST OF ABBREVIATIONS ............................................................................................................. 10
ABSTRACT ...................................................................................................................................... 11

CHAPTER

1 INTRODUCTION .......................................................................................................................... 13

1.1 Power Transistor Technologies .............................................................................................. 13
1.2 AlGaN/GaN HEMT Processing .................................................................................................. 17
1.3 AlGaN/GaN HEMT Operation .................................................................................................. 20
1.4 Factors Affecting AlGaN/GaN HEMT Electrical Characteristics ............................................ 24

2 EXPERIMENTAL METHODS ........................................................................................................ 31

2.1 Electrical Measurements .......................................................................................................... 32
2.2 Transmission Electron Microscopy .......................................................................................... 34
2.3 AlGaN/GaN HEMT Maskset and Chemical Deprocessing ......................................................... 43
2.4 Scanning Electron Microscopy ................................................................................................ 50
2.5 Surface Probe Microscopy ........................................................................................................ 56

3 OHMIC NANOCrack FORMATION AND ITS IMPACT ON RELIABILITY ......................... 68

3.1 Progress in Ohmic Contact Annealing ...................................................................................... 70
3.2 Metal Inclusions and Nanocrack Formation .......................................................................... 74
3.3 Nanocrack Morphology .......................................................................................................... 76

4 PROGRESS IN THE ANALYSIS OF THE RELIABILITY OF THE GATE ELECTRODE .......... 88

4.1 Inverse Piezoelectric Strain and the Reliability of Pt-Gate HEMTs ............................................ 89
4.2 Reliability of Ni-Gate HEMTs .................................................................................................. 98
4.3 Efforts in Surface Characterization of AlGaN/GaN HEMTs .................................................... 104

5 RELIABILITY OF THE GATE CONTACT .................................................................................. 112

5.1 Device Quality Below the Critical Voltage ............................................................................ 115
5.2 Defect Formation at the Critical Voltage .............................................................................. 116
5.3 Defect Formation Beyond the Critical Voltage ....................................................................... 128

6 CONCLUSIONS .......................................................................................................................... 145
<table>
<thead>
<tr>
<th>Section</th>
<th>Page</th>
</tr>
</thead>
<tbody>
<tr>
<td>LIST OF REFERENCES</td>
<td>150</td>
</tr>
<tr>
<td>BIOGRAPHICAL SKETCH</td>
<td>159</td>
</tr>
</tbody>
</table>
# LIST OF FIGURES

<table>
<thead>
<tr>
<th>Figure</th>
<th>Description</th>
<th>Page</th>
</tr>
</thead>
<tbody>
<tr>
<td>1-1</td>
<td>Schematics of two different power device technologies.</td>
<td>27</td>
</tr>
<tr>
<td>1-2</td>
<td>A performance diagram for a variety of power transistor families. Performance is gauged in terms of breakdown voltage and operational frequency.</td>
<td>27</td>
</tr>
<tr>
<td>1-3</td>
<td>A process flow for the creation of an AlGaN/GaN HEMT.</td>
<td>28</td>
</tr>
<tr>
<td>1-4</td>
<td>The electrical behavior of a schottky gate in forward and reversed bias.</td>
<td>29</td>
</tr>
<tr>
<td>1-5</td>
<td>A schematic of the operational modes of a transistor.</td>
<td>30</td>
</tr>
<tr>
<td>2-1</td>
<td>A schematic view of the electron optics utilized in a typical TEM.</td>
<td>61</td>
</tr>
<tr>
<td>2-2</td>
<td>The steps associated with Focused Ion Beam milling of a generic sample. Progression runs down the left column and then down the right column.</td>
<td>62</td>
</tr>
<tr>
<td>2-3</td>
<td>A top-down image of a typical maskset analyzed as part of this work.</td>
<td>63</td>
</tr>
<tr>
<td>2-4</td>
<td>The Deprocessing of an AlGaN/GaN HEMT.</td>
<td>63</td>
</tr>
<tr>
<td>2-5</td>
<td>AlGaN surfaces after exposure to HF.</td>
<td>64</td>
</tr>
<tr>
<td>2-6</td>
<td>AlGaN surfaces after exposure to various organic solvents.</td>
<td>65</td>
</tr>
<tr>
<td>2-7</td>
<td>Incomplete Removal of SiN from a T-gate.</td>
<td>66</td>
</tr>
<tr>
<td>2-8</td>
<td>HAADF-STEM images of deprocessing of the near-gate region of a HEMT.</td>
<td>66</td>
</tr>
<tr>
<td>2-9</td>
<td>A schematic diagram of a basic SEM system.</td>
<td>67</td>
</tr>
<tr>
<td>2-10</td>
<td>A schematic diagram of a generalized SPM system.</td>
<td>67</td>
</tr>
<tr>
<td>3-1</td>
<td>A HAADF-STEM image of the cross-section of a 100nm gate length device.</td>
<td>83</td>
</tr>
<tr>
<td>3-2</td>
<td>A metal inclusion formed after the annealing of a Al/Ni/Ti/Au metal stack.</td>
<td>83</td>
</tr>
<tr>
<td>3-3</td>
<td>Top-down SEM analysis of the ohmic contact regions of an AlGaN/GaN HEMT. These ohmic contacts were formed via an anneal at 850ºC for 30s.</td>
<td>84</td>
</tr>
<tr>
<td>3-4</td>
<td>FIB/TEM of a nanocrack observed in SEM.</td>
<td>84</td>
</tr>
<tr>
<td>3-5</td>
<td>Histograms of crack lengths observed in 20 HEMT devices.</td>
<td>85</td>
</tr>
<tr>
<td>3-6</td>
<td>Stepped stressing of an AlGaN/GaN HEMT and resulting crack formation.</td>
<td>86</td>
</tr>
<tr>
<td>Page</td>
<td>Description</td>
<td></td>
</tr>
<tr>
<td>------</td>
<td>-------------</td>
<td></td>
</tr>
<tr>
<td>3-7</td>
<td>The resulting crack distribution and associated tensile stress for crack growth in the stressed HEMT.</td>
<td></td>
</tr>
<tr>
<td>4-1</td>
<td>Piezoelectric strain (white arrows) resulting from a vertically applied electric field (black line) in an AlGaN/GaN HEMT.</td>
<td></td>
</tr>
<tr>
<td>4-2</td>
<td>The dependence of gate leakage current with increasing electrostatic stress.</td>
<td></td>
</tr>
<tr>
<td>4-3</td>
<td>Pulsed stress experiments on an AlGaN/GaN HEMT.</td>
<td></td>
</tr>
<tr>
<td>4-4</td>
<td>A schematic image of a crack-like defect formed at the drain side of the gate.</td>
<td></td>
</tr>
<tr>
<td>4-5</td>
<td>The transient electrical measurement performed by Kubal and coworkers.</td>
<td></td>
</tr>
<tr>
<td>4-6</td>
<td>An exemplary stressing experiment performed on a Ni-gated AlGaN/GaN HEMT.</td>
<td></td>
</tr>
<tr>
<td>4-7</td>
<td>BF-TEM image of a defect under the gate of an AlGaN/GaN HEMT.</td>
<td></td>
</tr>
<tr>
<td>5-1</td>
<td>A plot of gate current degradation and associated defect formation prior to ( V_{\text{CRIT}} ).</td>
<td></td>
</tr>
<tr>
<td>5-2</td>
<td>A plot of the gate degradation up to a voltage approximating ( V_{\text{CRIT}} ).</td>
<td></td>
</tr>
<tr>
<td>5-3</td>
<td>A plot of the degradation of a circular HEMT, with the same &quot;banding&quot; defect.</td>
<td></td>
</tr>
<tr>
<td>5-4</td>
<td>Morphological differences in banding induced by electrostatic stress and annealing.</td>
<td></td>
</tr>
<tr>
<td>5-5</td>
<td>The gate leakage current before and after annealing at 500°C for 30 min.</td>
<td></td>
</tr>
<tr>
<td>5-6</td>
<td>SPM of an AlGaN/GaN HEMT annealed at 500°C for 30 min.</td>
<td></td>
</tr>
<tr>
<td>5-7</td>
<td>HAADF-TEM of the interface before and after annealing.</td>
<td></td>
</tr>
<tr>
<td>5-8</td>
<td>HRTEM and FFTs of the Ni/AlGaN interface before and after annealing.</td>
<td></td>
</tr>
<tr>
<td>5-9</td>
<td>HAADF-STEM and EELS of the Ni/AlGaN interface before and after annealing.</td>
<td></td>
</tr>
<tr>
<td>5-10</td>
<td>An EDS linescan of the Ni/AlGaN interface before and after annealing.</td>
<td></td>
</tr>
<tr>
<td>5-11</td>
<td>Stepped stressing of a device to voltages well in excess of ( V_{\text{CRIT}} ).</td>
<td></td>
</tr>
<tr>
<td>5-12</td>
<td>Percentage of gate contact area consumed by under-gate defects.</td>
<td></td>
</tr>
<tr>
<td>Abbreviation</td>
<td>Description</td>
<td></td>
</tr>
<tr>
<td>--------------</td>
<td>-------------</td>
<td></td>
</tr>
<tr>
<td>AlGaN/GaN</td>
<td>Aluminum Gallium Nitride / Gallium Nitride</td>
<td></td>
</tr>
<tr>
<td>HEMT</td>
<td>High Electron Mobility Transistor</td>
<td></td>
</tr>
<tr>
<td>RF</td>
<td>Radio Frequency</td>
<td></td>
</tr>
<tr>
<td>RC</td>
<td>Resistance – Capacitance</td>
<td></td>
</tr>
<tr>
<td>BJT</td>
<td>Bipolar Junction Transistor</td>
<td></td>
</tr>
<tr>
<td>MOSFET</td>
<td>Metal Oxide Semiconductor Field Effect Transistor</td>
<td></td>
</tr>
<tr>
<td>HBT</td>
<td>Heterojunction Bipolar Transistor</td>
<td></td>
</tr>
<tr>
<td>TEM</td>
<td>Transmission Electron Microscopy</td>
<td></td>
</tr>
<tr>
<td>SEM</td>
<td>Scanning Electron Microscopy</td>
<td></td>
</tr>
<tr>
<td>UHR</td>
<td>Ultra High Resolution</td>
<td></td>
</tr>
<tr>
<td>TLD</td>
<td>Through Lens Detector</td>
<td></td>
</tr>
<tr>
<td>FIB</td>
<td>Focused Ion Beam Miller</td>
<td></td>
</tr>
<tr>
<td>DF</td>
<td>Dark Field</td>
<td></td>
</tr>
<tr>
<td>BF</td>
<td>Bright Field</td>
<td></td>
</tr>
<tr>
<td>HAADF</td>
<td>High Angle Annular Dark Field</td>
<td></td>
</tr>
<tr>
<td>EDS</td>
<td>Energy Dispersive X-Ray Spectroscopy</td>
<td></td>
</tr>
<tr>
<td>EELS</td>
<td>Electron Energy Loss Spectroscopy</td>
<td></td>
</tr>
<tr>
<td>STEM</td>
<td>Scanning Tunneling Electron Microscopy</td>
<td></td>
</tr>
<tr>
<td>SPM</td>
<td>Surface Probe Microscopy</td>
<td></td>
</tr>
<tr>
<td>V_{DS}</td>
<td>Source-to-Drain Voltage</td>
<td></td>
</tr>
<tr>
<td>V_{GS}</td>
<td>Source-to-Gate Voltage</td>
<td></td>
</tr>
<tr>
<td>I_{G}</td>
<td>Gate Current</td>
<td></td>
</tr>
<tr>
<td>V_{CRIT}</td>
<td>The critical voltage at which current through the gate increases</td>
<td></td>
</tr>
</tbody>
</table>
STUDIES ON THE RELIABILITY OF NI-GATE ALUMINUM GALLIUM NITRIDE / GALLIUM NITRIDE HIGH ELECTRON MOBILITY TRANSISTORS USING CHEMICAL DEPROCESSING

By

Patrick Guzek Whiting

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Aluminum Gallium Nitride / Gallium Nitride High Electron Mobility Transistors are becoming the technology of choice for applications where hundreds of volts need to be applied in a circuit at frequencies in the hundreds of gigahertz, such as microwave communications. However, because these devices are very new, their reliability in the field is not well understood, partly because of the stochastic nature of the defects which form as a result of their operation. Many analytical techniques are not well suited to the analysis of these defects because they sample regions of the device which are either too small or too large for accurate observation.

The use of chemical deprocessing in addition to surface-sensitive analysis techniques such as Scanning Electron Microscopy and Scanning Probe Microscopy can be utilized in the analysis of defect formation in devices formed with nickel gates. Hydrofluoric acid is used to etch the passivation nitride which covers the semiconducting layer of the transistor. A metal etch utilizing FeCN/KI is used to etch the ohmic and gate contacts of the device and a long exposure in various solvent solutions is used to remove organic contaminants, exposing the surface of the semiconducting layer for analysis.
Deprocessing was used in conjunction with a variety of metrology techniques to analyze three different defects. One of these defects is a nanoscale crack which emanates from metal inclusions formed during alloying of the ohmic contacts of the device prior to use in the field, could impact the yield of production-level manufacturing of these devices. This defect also appears to grow, in some cases, during electrostatic stressing. Another defect, a native oxide at the surface of the semiconducting layer which appears to react in the presence of an electric field, has not been observed before during post-mortem analysis of degraded devices. It could play a major part in the degradation of the gate contact during high-field, off-mode electrostatic stressing and could be the initiator of the pitting of the semiconducting layer of the gate contact, a defect which was also observed.
CHAPTER 1
INTRODUCTION

Aluminum Gallium Nitride / Gallium Nitride High Electron Mobility Transistors (AlGaN/GaN HEMTs) represent an emerging technology poised to greatly expand the application of power electronics, and specifically Monolithically Integrated Microwave Circuit technologies (MIMCs), in the field of Resonant Frequency (RF) and power electronics. Because of their status as an emerging technology in a manufacturing setting, demonstrating long term reliability of these devices remains a technical challenge.

In this chapter, an overview of power and microwave transistor technologies will be presented. It will be followed by a section explaining, in some detail, the process flow associated with the fabrication of an AlGaN/GaN HEMT, with an expository figure detailing the fabrication of the HEMTs to be used in this study. A section will be devoted to an overview of the device operation and a final section will be devoted to materials effects which influence operation.

1.1 Power Transistor Technologies

As wireless technologies become more and more ubiquitous, demand will increase for semiconductor devices which can be used in RF electronics applications. For a transistor technology to be useful in RF electronics, it must possess two separate figures of merit which compare favorably to other technologies. Firstly, a transistor used in RF applications must be capable of achieving high power densities. The power density is related to maximum voltage-current product achievable by the transistor. This power density is heavily influenced by the Breakdown Voltage, which is the voltage at which the transistor becomes incapable of carrying current while maintaining its ability to shut off current passing through the device. The Breakdown Voltage is, itself, heavily influenced by the difference in energy between the position
of the conduction band and the position of the valence band in the semiconducting material used in the fabrication of the transistor. Larger bandgaps lead to an increased dielectric constant for a given material, which reduces the electric field required to sustain a given voltage applied to the device's electrodes and, correspondingly, increases the maximum achievable voltage before dielectric breakdown occurs. Larger energy gaps between the conduction and valence bands of a semiconducting material also increase the voltage required to induce thermionic effects such as avalanche and gate leakage in a semiconductor devices [1].

Secondly, a transistor technology must be capable of reaching switching speeds which allow it to operate in the frequency bands useful for RF communications. The maximum frequency a transistor is capable of achieving is measured by considering the switching speed at which the gain of the transistor reaches a value of unity, also known as the threshold frequency [2]. The performance of a transistor at this frequency is dictated by the device architecture being used as well as a variety of other physical characteristics. The fundamental physical limit to a transistor's operating frequency for a given architecture is determined by the mobility and, with it, the saturation velocity of free carriers within the semiconducting current-carrying layers of the device. Higher free carrier mobilities and saturation velocities impart faster switching speeds. This is caused by an increase in the material conductivity and reductions in the associated critical Resistance-Capacitance (RC) time constants associated with operation.

GaN enjoys a substantial saturation velocity (25x10^6 cm/s) in comparison to a variety of other elemental and compound semiconductors (9x10^6 cm/s for Si, 9x10^6 cm/s for InP and 6x10^6 cm/s for GaAs) [1]. The implication of this is that, even though the electron mobility in the current-conducting region of a typical device is comparable to the mobility other material systems, AlGaN/GaN HEMTs are still capable of very high operating frequencies [3]. GaN and
AlGaN are very wide bandgap semiconductors, making them more akin to a semi-insulating material systems. The substantial bandgap observed in GaN and AlGaN is due to their crystal structure. When in thermal equilibrium, both materials form in a Wurtzite crystal structure characterized by alternating layers of Gallium (and Aluminum, when present) as well as Nitrogen. This forms a highly polar crystal structure where electrons are tightly bonded to their parent atoms [4]. The result are materials which are highly unreactive [5], with very large bandgaps [6]. Pure GaN has a bandgap of 3.1eV, while AlGaN has a bandgap which varies depending on stoichiometry from 3.1eV for GaN to 6.2eV for AlN. Because these materials have substantial bandgaps, they possess few intrinsic carriers, even at high temperatures, making them ideal for high temperature applications [7]. Two separate transistor architectures are commonly adopted for high frequency and high power applications. Bipolar Junction Transistors used in high power applications are three terminal devices in the "NPN" configuration. The emitter and collector of the BJT are doped with electron donors in order to create two separate electron-rich semiconductor regions. The collector and emitter are separated by a thin region of electron acceptor doped, hole-rich semiconductor material, called the Base. Because the emitter is doped with more electron donors than the collector, a diffusion current exists between the Emitter and the Collector. This current can be modulated by modulating the differential electron concentration between the emitter and collector. Larger differentials result in larger currents. The current in the transistor may also be modulated by modulating the current into the base and, as a result the hole concentration within the base [8].

Field-driven power devices function by forming a conducting channel between two electrodes: the Source (which supplies electrons) and Drain (which collects electrons). The resistance of the conducting channel between the Source and Drain of the device is modulated
electrostatically by the gate electrode, which is, ideally, electrically isolated from the channel. In the case of MOSFETs used in digital logic applications, this electrical isolation is achieved through the deposition or growth of an insulating layer; often SiO$_2$ or some metal oxide in silicon-based devices. In the case of HEMTs, which are always formed from compound semiconductors grown via epitaxy, this isolation is achieved by forming a schottky contact [9].

Current can be modulated in a field-driven transistor by varying the voltages on the source and drain electrodes because the conducting channel formed between the two electrodes acts as a resistor. Current can also be modulated by modulating the voltage on the gate to induce image charge in the conducting channel.

Most conventional field-driven devices utilize a conducting channel which is hole-rich, inducing free carrier depletion between the source and the channel. As a result, a voltage must be applied to the gate in order to induce inversion of the channel region to an electron-rich, conducting state. Because a voltage must be applied to the gate to induce a current, these are called Enhancement-Mode devices. Some semiconductors can't be doped to be hole-rich, however. So, no free-carrier depletion occurs between the Source and the channel. Field-driven devices formed from these systems must be inverted to a hole-rich state by the Gate in order to reduce current flow, making them Depletion Mode devices [10].

As shown in Figure 1-2, modern power transistor technologies cover a wide range of power and frequency applications. Silicon has been leveraged to great effect in applications where high frequency operation is required. Alloying with germanium to form silicon-germanium based Heterojunction Bipolar Transistors (HBT) has extended the frequency range accessible by silicon technology, but at the expense of breakdown voltage. This trade-off
between stability at high input voltages and high frequency operation is a general trend in high power electronics.

Indium phosphide (InP) and gallium arsenide (GaAs) are compound semiconductors which possess larger bandgaps and higher electron mobilities than silicon [11,12]. As a result, HBTs formed from these compounds generally demonstrate larger Breakdown Voltages and higher operational frequencies than HBTs formed from silicon or silicon-germanium alloys. However, the demand exists for devices with ever-higher Breakdown Voltages [13].

The large bandgap of GaN and AlGaN has limited their implementation in transistor technologies. The materials suffer from very low thermal conductivities, which makes heat conduction in AlGaN/GaN transistor technologies a significant limiting factor in their operation. Additionally, formation of hole-rich AlGaN and GaN layers has been difficult due to a lack of a useful electron acceptor to be used as a dopant [14]. Because of this, field effect devices such as HEMTs must be used in AlGaN/GaN systems rather than bipolar technologies such as BJTs or HBTs, which are faster than HEMTs by virtue of using a much smaller pathway for conduction. Because of their large bandgaps, HEMTs are capable of sustaining very large electric fields without breaking down, making them very attractive for high power applications. Improvements in material quality and device architectures are driving these devices towards higher threshold frequencies, as well.

1.2 AlGaN/GaN HEMT Processing

Epitaxy is required for the production of most GaN-based devices. The production of an AlGaN/GaN HEMT begins with the surface preparation of some substrate for epitaxial growth of GaN on its surface. In general, this seed crystal needs to be electrically insulating to prevent leakage through the bottom of the device but thermally conducting in order to mitigate self-heating of the HEMT. Seed crystals must also possess an atomic density and spacing similar to
that present in the GaN or AlGaN basal plane of the deposited semiconductor in order to reduce
defects associated with lattice mismatches. Three separate materials are typically employed for
this growth. 6H Silicon Carbide (SiC) [15], Sapphire [16] and (111) Silicon [17] are the most
common substrates used for these reasons. Sapphire was the first substrate used, because of its
hexagonal crystal structure and relatively high availability, but was supplanted by SiC when
thermal conductivity became a major factor in the operation of these devices. The following
procedure for the creation of an AlGaN HEMT is outlined and illustrated in Figure 1-2. This
figure highlights key components of the HEMT and also includes the processing conditions used
in the fabrication of HEMTs which will be used in this study [18].

The initial growth of GaN in modern technologies is generally performed on the surface
treated substrate via Metallorganic Chemical Vapor Deposition (MOCVD). On SiC substrates, a
nucleation layer of Aluminum Nitride (AlN) is employed to aid in epitaxial growth and to reduce
conduction through the substrate. A buffer layer of GaN is grown on the substrate via Metal-
Organic Chemical Vapor Deposition (MOCVD) to a thickness of, generally, a few microns.
This buffer layer is grown to a substantial thickness in order to trap defects and strain fields
formed due to lattice and thermal expansion coefficient mismatches between the GaN and the
SiC substrate. This layer is often intentionally doped with iron because it acts as a deep-level
electron donor, making it a source of n-type doping [19]. After the GaN buffer layer is grown,
an additional layer of AlGaN, with a 28% Al concentration in the case of this research, is grown
via MOCVD, which is capable of providing a lattice matched layer with minimal epitaxial stress.
The thickness of this layer is chosen to reduce tunneling current through the AlGaN to
acceptable levels while maximizing transconductance. It is made as thin as possible – in this
case, approximately 25 nm [18].
Even with the benefits of gettering and epitaxial growth, GaN/AlGaN layers are still rich in threading dislocations, which are present in state of the art material at concentrations of approximately $5 \times 10^9$/cm$^2$ [20]. Both of these layers are left only unintentionally doped in order to reduce the effects of impurity scattering on mobility.

Because of the highly polar nature of their bonding, AlGaN and GaN are materials which spontaneously form a constant electric field within their matrices and, correspondingly, a spontaneous polarization-induced charge at their interfaces. This field is further modified by the epitaxial strain on each layer due to the piezoelectric effect. When AlGaN is grown on top of GaN, the differential electric field results in a charged layer which is generally a few nanometers thick at the interface of the two materials. This layer of charge is called the Two Dimensional Electron Gas (2DEG) because of this tight confinement of charge along the c-axis of the AlGaN/GaN crystal which manifests itself in conduction band bending below the Fermi-Level [21]. The voltage at which the transistor switches on and off can be modulated by modulating the thickness of the AlGaN layer. After the deposition of AlGaN, a GaN-based cap layer is grown onto the epitaxial stack to reduce its reactivity with the ambient.

Access to the 2DEG for charge conduction is achieved by the deposition of and patterning Ti/Al/Ni/Au contacts, which form the Source and Drain electrodes of the transistor. A Rapid Thermal Anneal is performed after deposition and patterning in order to improve charge conduction out of the semiconductor and through these contacts. Annealing is supposed to achieve this through a process of metal diffusion down threading dislocations through the AlGaN and into the GaN. This brings the Source and Drain into intimate contact with the 2DEG, improving the ohmic nature of conduction through the HEMT [22].
After annealing is performed on the ohmic contact, the gate contact is formed through a combination of photolithographic lift-off patterning and metal deposition. The metal used for contact to the AlGaN surface must possess a workfunction which makes it an effective rectifying contact in order to limit gate leakage. In this work, nickel is used as a 20 nm liner layer which forms the Schottky contact and is capped with gold.

After the deposition of the gate metal, a capping layer formed from a dielectric oxide or nitride is deposited to reduce dangling bonds at the surface of the epitaxial stack, which tend to reduce the conductivity of the electron gas. At this point, the transistor is ready to undergo the processing for whatever interconnect technology and packaging is required for utilization. AlGaN/GaN HEMTs are used as discrete packaged devices and, more frequently, in Monolithic Microwave Integrated Circuit (MMIC) designs, where multiple metal interconnect layers are utilized.

1.3 AlGaN/GaN HEMT Operation

AlGaN/GaN HEMTs consist of a conducting two-dimensional sheet of electrons (the 2DEG) and a gate electrode which is isolated from the 2DEG and modulates its resistivity via electrostatics. The gate electrode is isolated from the 2DEG by the rectifying contact formed between the AlGaN and the gate.

This rectifying contact is a schottky diode formed between the metal of the gate contact and the AlGaN semiconductor. When these two materials are brought into intimate electrical contact, they must necessarily share the same fermi energy. However, they must also share the same vacuum level. To satisfy both of these requirements, the valence and conduction bands of the semiconductor must bend near the interface between the metal and the semiconductor. The direction in which the semiconductor's bands bend is dependant upon where the fermi levels of the semiconductor and metal sit in relation to the vacuum level. The difference in energy
between the fermi level of the metal or of the semiconductor and the vacuum level is called that material's workfunction.

If the workfunction of the metal is larger than the workfunction of the semiconductor, the conduction and valence bands will bend up to allow the two materials to equilibrate when they are joined. If the workfunction of the metal is smaller than the workfunction of the semiconductor, the conduction and valence bands will bend down to allow the two materials to equilibrate. This can be represented by the built in voltage, $V_{BI}$ (in volts), which represents the turn on voltage for the schottky diode and may be expressed as the difference between the metal workfunction, $\Phi_M$ (in eV), and the semiconductor workfunction, $\Phi_S$ (in eV), divided by a fundamental electronic charge [23].

Epitaxial AlGaN used in HEMT gate stacks is generally only unintentionally doped, which makes the difference between the conduction band energy and the fermi energy roughly equal to half the distance between the valence and conduction bands. In most cases, intrinsic AlGaN is slightly n-type in the as-deposited state. In n-type materials, a rectifying contact will induce depletion at the surface between the metal and the semiconductor, resulting in hole accumulation near the interface. This occurs when the metal workfunction is larger than that of the semiconductor and the conduction and valence bands bend down.

Given that depletion occurs, it is not surprising that the current equation for a gate contact is similar to that for a pn junction diode. The current through the diode, $I_G$ (in A), is as follows,

$$I_G = A J_{RG}$$

where $A$ is the contact area (in cm$^2$), $J_{RG}$ is the recombination-generation current density and $A^*$ is Richardson's Constant (in A/cm$^2$-K$^2$). The difference in the metal and semiconductor workfunctions may be expressed as a built in voltage, $V_{BI}$, equal to the difference between the metal workfunction and the semiconductor workfunction divided by the charge of a single
electron. $V_G$ is the potential on the gate while $V_{2DEG}$ is the potential applied to the 2DEG (in volts), which may vary along the length of the gate electrode [24].

$$I_G = A\left(J_{RG} + A'T^2 e^{V_{GS} - V_{2DEG} - V_E - 1}\right) \quad (1-1)$$

As shown in Figure 1-5, the gate contact electrostatically modulates the resistivity of the 2DEG which is situated directly below the gate contact and is, generally, well isolated by the AlGaN from the 2DEG as long as the gate is not forward biased. In this way, the conductive channel of the gate may be turned on and off by placing a potential on the gate electrode, inducing image charges within the 2DEG which increase or decrease the carrier concentration.

The threshold voltage of this device is induced by charges accumulated at the interfaces between the GaN and AlGaN as well as the AlGaN and the gate metal. It can be expressed as a sum of potentials generated by these charged interfaces with the following equation, where $\sigma_{PZ}$ is the spontaneous polarization and inverse-piezoelectric effect induced charge at the AlGaN/GaN interface (in cm$^{-2}$), $t_{AlGaN}$ is the thickness of the AlGaN layer, $\varepsilon_{AlGaN}$ is the permittivity of the AlGaN, and $\Delta E_C$ is the conduction band offset (in eV) between the AlGaN and GaN. The assumption is made that the AlGaN layer is undoped [25].

$$V_{TH} = V_{BI} + E_F + \Delta E_C - \left(t_{AlGaN} \sigma_{PZ} / \varepsilon_{AlGaN}\right) \quad (1-2)$$

If $V_{GS}$ doesn’t exceed the threshold voltage, the section of the 2DEG under the gate electrode will be depleted of charge carriers, and the 2DEG will, ideally, not pass any current regardless of the potential applied to the drain. This is known as "off" mode. As the potential on the gate exceeds the threshold voltage, the transistor switches into the linear regime, or "semi-on" mode, as it is described by some authors. In this regime, the current flow through the device increases roughly linearly with increasing potential on the drain. The extent of this linear regime is defined by the following condition, where $V_{GS}$ represents the gate-to-source voltage and $V_{TH}$
represents the threshold voltage of the device, and $V_{DS}$ represents the drain-to-source voltage [26].

$$V_{DS} < V_{GS} - V_{TH}$$  \hspace{1cm} (1-3)

If the above condition is not met, the transistor enters the saturation regime, where additional voltage does not result in any additional current flow. This occurs because the part of the channel close to the drain side of the device remains depleted, or "pinched off" and the transistor self-limits.

In order to calculate the current flowing through the channel of the AlGaN/GaN HEMT, the electrostatics of the device must be considered as the gate voltage is raised above $V_{TH}$, and the transistor is switched out of "off" mode in order to form a conducting channel. The current density within the device at any point along the channel is equal to the conductivity of the channel multiplied by the lateral electric field. The conductivity of the channel is dependant upon such variables as the fundamental charge of an electron ($q$), the mobility ($\mu$), as well as the carrier concentration. The carrier concentration is derived from the capacitance of the AlGaN (which is, itself, equal to $\varepsilon_{AlGaN}$ divided by $t_{AlGaN}$) as well as the potential difference between $V_{GS}$ and $V_{TH}$. The electric field is equal to the change in $V_{2DEG}$ with respect to position along the channel as a consequence of potential applied to the drain electrode of the device.

In order to determine the current passing through the full device, rather than through a single infinitessimal slice thereof, this current density equation is integrated over the full width, $W$ (in cm), and full length, $L$ (in cm), of the device. This yields the relationship between the current flowing from the source electrode to the drain electrode and the voltages applied to the various electrodes present within the HEMT, but only for the linear regime of operation.
(i.e. where the channel has not yet been "pinched off"). The constant saturation current may be calculated by applying the constraint of the condition listed in Equation 1-3 to Equation 1-4 [27].

\[ I = \left[ q\mu \left( W/L \right) \left( \varepsilon_{\text{AlGaN}} / \varepsilon_{\text{AlGaN}} \right) \right] \left[ (V_{GS} - V_{TR}) (V_{DS}) (V_{DS}^2 / 2) \right] \]  

(1-4)

1.4 Factors Affecting AlGaN/GaN HEMT Electrical Characteristics

As is evidenced by the previous section, the conduction of charge carriers through an AlGaN/GaN HEMT is influenced by a variety of factors, some of which are dependant upon geometry (such as the length and width of the HEMT, the thickness of the AlGaN layer, etc.). However, a variety of materials parameters also affect ideal device operation. Their variability both during processing and after processing, in the field, are of prime importance.

It should also be noted that geometry also influences the characteristics of the HEMT, especially at short length scales, where short channel effects such as velocity saturation and hot electrons cause the device to perform differently than what would be expected given the equations outlined in the above section. To a first order, however, the equations above and their dependance upon given materials constants is accurate.

Saturation current is affected by several material characteristics. The dielectric constant of the AlGaN, which is dependant upon the stoichiometric composition of the epitaxial layer, has a direct influence upon the current observed passing through the device. Likewise, the carrier mobility within the 2DEG also directly influences the magnitude of saturation current observed passing through the channel.

A variety of factors can influence the carrier mobility in the 2DEG, the most commonly encountered being scattering by impurities or trap centers. Trap centers can arise from a variety of sources in an AlGaN/GaN HEMT. Dangling bonds found at unpassivated or imperfectly passivated surfaces (such as the top surface of the AlGaN) can form deep level traps which act as scattering centers which reduce the mobility of carriers in the 2DEG. This phenomenon is of
particular importance in the field of device reliability, where stressing involving high fields (and hot electrons) can greatly impact the concentration of trap centers present on passivated and unpassivated surfaces alike [28]. A variety of passivation layers have been explored for use in reducing surface states in AlGaN/GaN HEMTs, from SiN [29], to Al₂O₃[30], to Sc₂O₃[31]. Threading dislocations and misfit dislocations, which are a nearly-unavoidable consequence of epitaxial growth of GaN beyond its pseudomorphic thickness, also act as scattering and non-radiative recombination centers which influence carrier mobility and minority carrier lifetimes [32]. Even the iron doping used to improve the quality of MOCVD-grown GaN can have a deleterious effect on the carrier dynamics associated with device operation because they act as deep level electron acceptor [33].

It bears mentioning that these ionized scattering centers can also compensate the charge levels within the 2DEG of a HEMT, further reducing the conductivity of the channel region. This effect is often observed as an increase in the "access" resistances associated with the source-drain contacts of the device. These access resistances are not captured in the physics of operation described in the previous section, but they can be considered as connected in series with the transistining element of the HEMT which was described in the previous section. Increases in access resistance also manifest themselves as a reduction in the observed saturation current in the HEMT device, despite the fact that their effect is not directly captured by the current equations which were outlined previously.

The ideality of the Schottky diode utilized in the gate contact of the HEMT is also of critical importance to its proper operation. The ability of the contact to sustain low reverse-biased currents, even at high applied fields, ensures that leakage from the 2DEG of the device and the resulting power dissipation is minimized in "off" mode. As before, trap centers present
at surfaces as well as within the bulk of the AlGaN play a large role in the quality of this contact in reverse-biased mode. These trap centers can induce recombination-generation current, which increases the current observed in the diode in the reverse-biased mode of operation. These trap centers can also induce fermi level pinning in the device, which influences the observed energetic barrier to current conduction, and associated turn-on voltage of the gate contact [26]. Reactions between the gate contact and the underlying AlGaN and GaN can also generate parallel paths of conduction from the 2DEG and the gate electrode, effectively shorting out the schottky contact and greatly increasing the observed leakage current in the device.

The reliability of AlGaN/GaN HEMTs studied as part of this work is influenced by all of these materials factors. Degradation of the electrical properties of the device can be linked back to changes in the structure of the device and associated changes in the materials constants which influence the ideal behavior of the HEMT. It will be the purpose of this work to outline the effects of structural abnormalities in HEMTs, observed with various microscopy techniques both before and after electrical stressing, on the electrical properties of these devices.
Figure 1-1. Schematics of two different power device technologies. A) A simple operational schematic describing the structure of a BJT, which operates on diffusion current, is shown. B) The structure of a High Electron Mobility Transistor, which is a field effect device, is shown.

Figure 1-2. A performance diagram for a variety of power transistor families. Performance is gauged in terms of breakdown voltage and operational frequency. Included are Si and SiGe BJTs and HBTs, InP HBTs and HEMTs, GaAs HBTs and AlGaN/GaN HEMTs. The tendency towards reduced breakdown voltage at higher frequency is apparent, as is the trend of increasing breakdown voltage with increasing band-gap [1].
Figure 1-3  A process flow for the creation of an AlGaN/GaN HEMT [19].
Figure 1-4  The electrical behavior of a schottky gate in forward and reversed bias.  A)  A diagram showing the band bending which occurs both at the instant when a metal and n-type semiconductor are joined and when the fermi level (equal in this case to the mid-gap energy) has a time to equilibrate.  B) A diagram of the forward biased (positive voltage) and reverse biased (negative voltage) currents with respect to the potential applied to the gate.
Figure 1-5 A schematic of the operational modes of a transistor. When the potential supplied to the gate is smaller than the threshold voltage, $V_{TH}$, the 2DEG is depleted and the transistor is in "Off" mode. As the gate voltage exceeds the threshold voltage, the transistor can either enter the linear or saturation regimes, depending upon if $V_{DS}$ is greater than the difference between $V_{GS}$ and $V_{TH}$. If $V_{DS}$ is larger than $V_{GS}-V_{TH}$, the channel pinches off near the drain side of the device and the transistor saturates.
CHAPTER 2
EXPERIMENTAL METHODS

A variety of techniques are used in analysis of semiconductor device reliability. The techniques utilized in reliability studies cover a wide spectrum of electrical and structural materials properties. Metrology techniques can also be used to identify processing-related defects within a device which do not directly correlate to materials properties as well as to identify poor device operation (i.e. poor electrical behavior).

This section of the dissertation will focus on the analytical methods used to characterize the AlGaN/GaN HEMTs used in these studies. It will begin with a discussion of two of the more-commonly utilized techniques in the literature. The first of these are the electrical measurements made before, after and (sometimes) during electrical stressing. These electrical measurements are the first step in the detection of defects induced by normal device operation and are generally used in conjunction with microscopic and spectroscopic techniques to correlate changes in electrical properties with changes in structural properties. Arguably the most popular technique, Transmission Electron Microscopy (TEM), will be discussed after the section on electrical measurements. As will be shown, this technique is of limited use in analysis of the stochastic processes which lead to electrical defect formation in these HEMTs.

These sections will be followed by a discussion of the chemical deprocessing scheme developed as part of this work and its utility in analyzing stochastic processes such as defect formation in the HEMTs used in this study. Some special attention will be paid to variations in the deprocessing scheme and their effects on sample quality. This section will be followed by a discussion of Scanning Electron Microscopy (SEM) and Atomic Force Microscopy (AFM). These techniques proved to be of great utility in the characterization of the surface of the AlGaN
epitaxial layers after deprocessing. Special attention will be paid to the settings used in both of these technique to provide high quality imaging of the AlGaN surface.

2.1 Electrical Measurements

Electrical stressing and electrical measurements of AlGaN/GaN HEMTs were performed on the same instrumentation, as is commonly the case in reliability studies because of the convenience of not having to switch the sample from one system to another to make measurements after stressing. This is especially true in cases where stepped stressing experiments are called for, where having to physically move the sample for each stressing step would greatly impede throughput. The system used for electrical analysis in all studies described in this work was an HP4146 parameter analyzer mated to a Tectronix 370A curve tracer.

These systems are capable of multiple parallel measurements of various parameters both before, during, and after stressing. Generally, analysis of a device under stress begins and ends with a cursory examination of its electrical characteristics. This examination begins with an analysis of the characteristics of the gate contact, where the gate-to-source current \( I_{GS} \) is plotted against the gate-to-source voltage \( V_{GS} \). Generally, this measurement is carried out from \( V_{GS} \) equal to 1V (the schottky diode being "on" in this case) out to \( V_{GS} \) equal to some significant negative voltage (where the diode is "off"), in 0.1 V increments. The ability of the gate to turn the 2DEG channel "on" and "off" (i.e. its transconductance) is also measured via a plot of the drain-to-source current \( I_{DS} \) versus the gate voltage \( V_{GS} \). This analysis generally occurs over the same voltage range as the measurement of the gate current and with 0.1 V increments on \( V_{GS} \), as described previously.

The behavior of the device in saturation is also of critical importance to understanding the overall quality and degradation of a device before and after stressing. To this end, the traditional "family of curves" plot common to most FETs is employed. This measurement involves analysis
of $I_{DS}$ as the voltage between the source and drain ($V_{DS}$) is swept from a value of 0 V to some value which induces saturation in the device (usually around 4 V) in 0.1V increments.

Generally, $V_{GS}$ is also stepped in order to generate a family of curves rather than a single curve and to observe the increase in $V_{DS}$ required to induce saturation in the device as $V_{GS}$ is increased. A typical range of values for $V_{GS}$ to form a family of curves is -3 V to 1 V stepped in 1 V increments.

Stressing of devices under analysis for this study took place in an air ambient under standard temperature and pressure. Stressing can involve maintenance of a device at a single voltage or in a "stepped" configuration where the voltage is incremented at a specified rate over a given period of time.

Currents flowing through any terminal of the HP4146C cannot exceed a level of 100 mA without causing damage to the testing system. Because of this, devices which were electrostatically stressed in this study were stressed in off-mode, where the channel of each device was pinched off with a large negative voltage on the gate electrode. This limits the leakage current flowing through the device to a level which does not induce compliance within the testing system.

Depending on the specifics of the study performed one voltage (either $V_{DS} \text{ or } V_{GS}$) would be held constant as another voltage (again, either $V_{DS} \text{ or } V_{GS}$) was stepped with an increment of -1 V/min. Generally, this stepped stressing continued until some maximum voltage was achieved or until the device under stress reached one of the preset current compliances applied to all electrodes. Stressing ended on devices which were stressed as part of this study when the gate contact reached compliance, which varied from study to study, ranging from 100 µA to 1 mA on
the gate contact of the device. The gate electrode was, consistently, the first electrode to achieve the current dictated by the set compliance.

During stressing, devices were subjected to a set of abbreviated measurements after each voltage increment. This measurement, generally, occurred over a 1min time period, resulting in a 50% duty cycle for stressing. These abbreviated measurements consisted of analysis of $I_{GS}$ at -1V $V_{GS}$ in order to determine the magnitude of leakage current through the gate electrode. The respective currents through both the gate electrode and the source/drain were recorded during stressing in order to determine leakage current in the high-field case as well.

It is important to point out that electrical measurements form the baseline for all studies into the reliability of the devices studied as part of this work. In order to successfully "diagnose" and remedy degradation and failure of devices, structural changes must be linked to changes in electrical properties. Without electrical measurements which demonstrate that the overall quality of the device under test changed as a result of stressing, observations of physical abnormalities in the structure of a stressed device can only yield educated guesses as to how the physics and chemistry of degradation in the field adversely effects device characteristics.

2.2 Transmission Electron Microscopy

Transmission Electron Microscopy (TEM) functions by forming a coherent source of electrons and accelerating these electrons through a large potential difference, through an electron transparent layer of sample material and into some electron sensitive system which acts as a sensor. The image which is collected by this sensor can yield detailed qualitative and quantitative data about the representative sample material. This data can range from information about the density and structure of the material under observation, to chemical information detailing atomic ratios and (in some cases) atomic bonding, and even to electric fields present within the sample (via techniques such as electron holography) [34]. Given the wealth of data
available from this technique, it should not be surprising that TEM represents one of the more oft-employed analytical techniques in the field of semiconductor reliability. The two TEMs utilized in this study were a JEOL-2010F and a JEOL-ARM200F.

A basic TEM consists of several different modules which work to generate the electron beam, direct that beam through a cross-sectioned sample, and collect the electrons which were transmitted through to the other side of that sample. The first of these modules is the electron source, which is used to extract the electron current used for imaging from a filamentary source by means of thermionic or field emission. Sources vary from TEM to TEM, but the most high performance of these sources is a field emission gun (FEG) source which can be used to extract electrons with minimal energetic variation and with a smaller starting probe size, making the entire microscope less prone to chromatic lens aberration. The extracted electrons exist at high potential and are accelerated through the microscope column, and towards lower potentials, as a result [35].

After the electrons are accelerated out of the source, they encounter a pair of electromagnetic quadrupole or hexapole lenses, known as the condenser lenses, which are used in conjunction with an aperture just above them in the column to form a coherent beam from the electron source image. In Bright Field TEM, the condenser lenses are adjusted to form a parallel beam to uniformly illuminate the sample. In the case of Scanning Transmission Electron Microscopy (STEM), the lenses are adjusted so that the electron beam is convergent on the sample, rather than parallel to the optic axis [36].

In some modern systems, such as the JEOL-ARM200F, additional lensing is installed in order to correct for spherical aberration in the beam optics [37]. After passing through these additional lens elements, the electrons encounter the cross-sectioned specimen. At this point, the
electrons may either pass through without interference, scatter off of the material which comprises the sample itself, or they may be absorbed.

The component of the incident beam which is scattered by atoms within the sample can scatter in two very different ways. If the region of the cross-sectioned sample which the electrons passed through is amorphous, there is no preferential direction associated with scattering, and the diffraction pattern formed assumes the appearance of a diffuse "cloud" (with a radius approximately equal to the nearest neighbor distance of the amorphous material) if it is imaged directly. In the case of a crystalline sample, electron diffraction occurs and the diffraction pattern occupies very specific positions, when imaged directly. If the beam passes through a region with many different crystal directions represented, the resulting diffracted pattern will form rings rather than the singular spots associated with diffraction in a single crystal.

After the electrons pass through the sample, they encounter the objective lens and, below it, its associated aperture. The lens collects electrons which comprised the parallel beam incident to the sample and focuses this beam on its back focal plane. It is easy to observe from any ray diagram of the system described above that, because the electron beam incident to the sample is parallel to the optic axis, every radial position of the objective lens collects diffraction information about the entire image of the sample. The objective aperture is used to screen out information which enters the objective lens at high angles, improving contrast and increasing the depth of field of the instrument.

The intermediate aperture is placed after the objective lens and is used to collect only certain diffraction information about the specimen. If the aperture is placed on-axis, the component of the electron beam which passes through the aperture transmits information only
about the mass-thickness variation data contained by the direct beam. This is called bright field imaging. If the aperture is placed off-axis (or the beam tilted so that the axis shifts), the transmitted beam contains diffraction information relating to electrons scattered only in the direction corresponding to the position of the opening in the intermediate aperture. This is called dark field imaging. Both bright field and dark field imaging were utilized in this work.

In order to collect the image formed by the transmission of electrons through the transparent sample, additional electromagnetic optics, in the form of the intermediate lens, are required in order to form an image on the projector screen, film, or CCD used to capture the image. This is accomplished by adjusting the power of the intermediate lens such that it is focused on the image plane of the objective lens. This results in the formation of either a bright field or dark field image on the projection plane of the microscope, depending upon the orientation of the intermediate aperture. If diffraction data is desired, this lens is weakened until it is focused on the back focal plane of the objective lens, rather than the image plane. In the case of diffraction, the relationship between where the characteristic bright spots (from a single crystal) or rings (from a polycrystalline sample) form in the final projected image of the diffraction data.

These positions are governed by the camera length equation, wherein the distance between the projection screen and the sample under analysis is defined by the variable \( L \), the distance between diffracting crystal planes is defined by the variable \( d \), the deBroglie wavelength of the incident electron is defined by the greek symbol \( \lambda \), and the distance from the bright field spot to the diffracted spot of interest is defined by the variable \( r \) [38].

\[
rd = \lambda L
\]  

(2-1)
The deBroglie wavelength of the electron is defined by the equation below, where $h$ represents the Planck constant, $V$ is the accelerating voltage, $m$ is the rest mass of the electron and $q$ is the fundamental charge of an electron [39].

$$\lambda = (-qV/hc)(1 + [-qV/2mc^2])^{-1/2}$$  \hspace{1cm} (2-2)

Thus, diffraction spots which are formed from small interplanar spacings have large displacements from the direct beam of the electron signal while diffractions spots which are formed from large interplanar spacing have small displacements from the direct beam. It is important to note that, while this equation governs the position of scattered electrons in selected area diffraction, it also provides the same information which is necessary in generating a dark field image using a given diffraction bright spot. It also has utility in the convergent beam imaging which is utilized as part of Scanning Transmission Electron Microscopy (STEM), which will be detailed in the following part of this section.

STEM is of particular utility in situations where high resolution imaging is required. These are generally situations where a feature must be imaged with dimensions approaching the deBroglie wavelength of the electrons accelerated from the source. STEM functions by increasing the power of the second condenser lens such that the beam crosses over, or converges, at the sample surface. This converged electron beams beam converts the direct spot and diffraction spots, typically generated during illumination of the sample with a parallel beam, into a group of cones corresponding to each spot. The radii of the projected images formed from these cones follow from the convergence angle of the beam while their positioning relative to one another is determined by the camera length equation, described above. The image resulting from one of these resultant "cones" of information is called a Convergent Beam Electron Diffraction (CBED) image [40].
The high resolution image generated by the TEM arises from the interference between the direct spot CBED image and the CBED images which result from the diffracted electrons. A proper STEM image is formed by rastering the converged beam across the sample surface and by recording the overall intensity of the resultant image collected by the detector. Because this intensity is impacted by the interference between the CBED images formed by diffraction of the convergent beam, it is not surprising that the image formed is dependant upon which CBED cones are collected by the detector in use and, by association, the type of detector which is inserted into the beamline.

Generally, STEM is performed with two different types of detectors. If the CBED image resulting from the direct beam is of primary importance, a centered CCD array is utilized in order to collect the transmitted CBED image and the interference pattern arising from the diffracting CBED images. The resulting information takes on the form of a bright field STEM image. If a dark-field image is preferred, an annular detector is favored over the centered CCD array. This detector collects the components CBED image displaced from the optical axis by a distance larger than the inner radius of the detector (i.e. any CBED image formed by scattering to an angle beyond the minimum collection angle of the annular detector). The resulting interference patterns detected by the annular ring form a dark field STEM image.

In general, high resolution images generated via TEM (such as STEM images) can be used to determine structural information about the material under analysis without the need for an image of the the electron diffraction pattern generated by the incident beam. This is accomplished by means of a Fast Fourier Transform (FFT), which is a matrix operation that converts the data contained within the high resolution image into a map of the component
frequencies associated with the repeating patterns contained within that image. An FFT treats these repeating patterns as a sum of sinusoidal functions [41].

STEM is also favorable over traditional TEM imaging techniques in cases where quantitative microscopy is desired in additional to high resolution imaging. The small footprint of the rastered beam makes it ideal for analysis of features at the nanoscale utilizing spectroscopic techniques such as X-Ray Energy Dispersive Spectroscopy (EDS) [42] or Electron Energy Loss Spectroscopy (EELS) [43]. While there are a variety of other techniques commonly used in conjunction with STEM (such as Cathodoluminescence [44] and Electron Holography, to name a couple more), EDS and EELS will be the techniques favored as part of this work.

As electrons enter the sample from the converged beam generated at the condenser lens, they may interact and scatter off of atoms present within this sample. This scattering is the result of inelastic collisions with the electron clouds of the sample's constituent atoms, where some of the energy of the incident electron is imparted to the electron shells of the atoms which it interacted with. In EELS, the energy lost by the electron in transit through the sample is determined through the use of an EELS detector placed in line with the optical axis of the microscope.

As electrons enter the EELS detector, they encounter a static electric field which bends them in such a way that the electrons separate in the field by energy before colliding with a CCD array. Electrons with low momentum and low kinetic energy are bent more readily than electrons with a greater momentum. Thus the variable intensity of the resultant electron beam with respect to position on the CCD array can be used to determine the proportional energy losses experienced by electrons being transmitted through a sample. The characteristic energy losses associated with transmission can be used to determine the chemistry of the material under
analysis. Atomic ratios and, in some cases, even bonding can be determined from a high quality EELS spectrum. Also, because only the electrons which undergo minimal scattering are utilized in this technique, the EELS detector can be combined with the annular ring detector described above to yield complementary EELS and dark field STEM imaging in tandem on a sample.

When incident electrons interact with the electron shells of the constituent atoms of the sample and lose energy, this imparted energy can result in a variety of interactions within those constituent atoms which the electrons collided with. If an incident electron knocks a core electron free, a valence electron will lose energy and take the place of the lost core electron. The energy lost by this valence electron exits the sample in the form of an x-ray, which can be collected by a photodiode chilled by liquid nitrogen. The energy of this x-ray can be determined by the number of photoelectrons which are collected by the diode as a result of stimulation by the incident photon. Not surprisingly, the energy of the incident photon is characteristic of the interorbital transition which the valence electron went through as it fell from one shell to another within the atom. If the incident photons collected by the photodiode are collated with respect to their energies, and EDS spectrum results and this EDS spectrum corresponds (albeit not directly due to matrixing effects which cannot be separated without a standard for comparison) to the relative ratios of different atoms present within a sample. In this way, EDS can yield substantial quantitative information regarding the chemistry of a material under analysis.

TEM is particularly effective when coupled with equipment which can extract a material sample from a specific site in a semiconductor device. This technique, called Focused Ion Beam (FIB) milling, utilizes a Scanning Electron Microscope with a secondary column of electron optics linked to a liquid gallium source which acts as a supply of gallium ions for site-selective ablation. FIB-based sample preparation makes it possible for an electron transparent sample to
be cut from a HEMT, mounted to a copper grid suitable for insertion into the electron column of a TEM, and thinned – all in situ and without breaking a vacuum [45]. This process is shown in Figure 2-2.

The first step in this process is to deposit a protective layer, generally over 100nm of thermally evaporated carbon, onto the surface of the sample. This occurs prior to insertion into the FIB. In samples where the structure itself is protective to the region of interest, this process can be skipped. The sample is then inserted into the FIB and a vacuum of $5 \times 10^{-5}$ mmHg achieved by means of a turbo pump. The sample is set at eucentric height by means of SEM imaging and is then tilted $52^\circ$ into the gallium beamline.

After insertion of a platinum GIS needle, site-selective deposition occurs over the area of interest as the gallium beam is rastered across a beam of organomettallic platinum molecules. This results in the deposition of a dense platinum layer, which acts as a protective mask in the subsequent milling of two trenches to either side of the site of interest with depths large enough to expose the region of interest under the deposited metal. The sample is thinned down by progressive milling steps to a thickness of approximately $1\mu$m and a cut is made which frees the bottom of the cross-sectioned region of interest from the larger sample. A second cut is made which frees one of the sides.

At this point, a second attachment known as an omniprobe needle is inserted into the beamline of the chamber and the sample is tilted parallel with the electron beamline, putting it at $52^\circ$ to the gallium beamline. For the purposes of in-situ lift out, this needle must be hard (so that it can withstand the physical abuse of making solid contact with a surface without deforming plastically), but also thin (to facilitate bonding the cross-section to the omniprobe needle). A tungsten needle thinned to a radius of curvature of $1\mu$m is perfect for this application. The needle
is brought into intimate contact with the cross-sectioned area of interest and the Pt source is used, again, to weld the top of the cross-sectioned sample to the omniprobe needle. At this time, the remaining attached side of the cross-section is cut free and the cross-section transported to the TEM grid. At this point, it is welded to the grid and the omniprobe needle is cut free.

At this point, both the Pt GIS attachment and the omniprobe attachment are retracted and they remain in this position. The mounted sample is tilted into the gallium beamline. The sample is now ready for final thinning. The final thinning steps begin with a group of successive milling steps which thin the sample down to a few hundred nanometers. At this point, the sample is often wedge cut, so as to taper the long side of the cross-section, from one side to the other, with the intent of generating an ideal sample for analysis close to the tip of the wedge. The sample is then tilted off axis from the gallium beam by four degrees and a mill is performed at a low accelerating voltage (7 kV) to clean the surface of the sample prior to analysis with TEM.

2.3 AlGaN/GaN HEMT Maskset and Chemical Deprocessing

It is important to note that, while TEM is a powerful analytical tool for site-specific high resolution imaging of nanoscale features, it suffers from a tremendous disadvantage in the analysis of stochastic processes, which comprise the majority of defects observed in AlGaN/GaN HEMTs. This disadvantage is the small total volume of any sample generated for TEM. In order to form an electron-transparent sample, a TEM cross-sectioned must be thinned to a thickness of, at maximum, 100nm. In fact, some estimates place the total volume of all TEM cross-sections created between 1955 and 1970 as being less than one cubic millimeter [46]. For this reason, caution must be used whenever a study generated by TEM cross-sectioning is being employed to understand the fundamental properties of a larger material set.
This has been found to be especially true for cross-sectioning of samples formed from AlGaN/GaN HEMTs, where defect formation is highly stochastic in nature. The morphology of a defect can vary dramatically depending upon where a cross-section is taken from a device. In many cases, a defect may be present in one area of a device, but not in another. In fact, some of the stochastic defects observed in this work only occur over a very small percentage (often less than 5%) of the total gate area and there have been many cases where a device demonstrated degradation electrically, but no TEM cross-sections yielded a sample with an observable defect present. This null result, which opposes observed electrical data, makes it difficult to derive a conclusion about the mechanism of defect formation from TEM data alone.

Because TEM cross-sectioning is a destructive process and because the defects formed under the gate during stressing of these devices cannot be directly imaged non-destructively prior to cross-sectioning, no means exists of determining the nature of a defect in a degraded HEMT prior to its being cross-sectioned and all cross-sectioning must be performed "blind", without any prior knowledge of whether or not a defect exists at all in a cross-sectioned volume and also without any prior knowledge of what that defect's morphology is if it is actually present in a given cross-sectioned area.

Because of this, TEM is not aptly suited for analyzing and quantifying the stochastic nature of defect formation in AlGaN/GaN HEMTs and other analytical techniques are required in order to analyze the progression of defect formation quantitatively. The methodology utilized in this study was deprocessing of the AlGaN/GaN HEMT such that the passivation nitride and metal layers are removed from the device structure. This exposes the AlGaN epitaxial layer of the device to direct analysis using top-down imaging techniques such as SEM and SPM, which will be described in greater detail later in this chapter. This deprocessing method has been
shown to be highly effective for use in analysis of defects formed at the surface of this AlGaN layer as well as for use in analysis of volumetric inclusions within the layer.

The deprocessing strategy employed varies slightly with the device under analysis. With this in mind, it seems beneficial to discuss, briefly, the maskset used as part of this study. As shown in Figure 2-3, the maskset used to generate the HEMT structures contains a variety of structures, including a group of spaced ohmic contact pads used for Transmission Line Measurements (TLM), a lithographic alignment vernier, and five separate HEMT structures. Of these HEMT structures, three are of particular interest because of their use in the studies detailed in this work. These are the device with a submicron-scale "T-gate" structure, a device with a 1µm gate length, and a device with a much larger, 60µm gate length (called the FatFET). All of these HEMT structures utilized nickel as the gate metal of choice.

Deprocessing begins with an exposure to hydrofluoric acid in the form of buffered oxide etch with a 6:1 stoichiometric ratio of HF to NH\textsubscript{3}F. HF etching occurs for a total of 15 minutes, during which time the PECVD silicon nitride which passivates the AlGaN surface of the HEMT is dissolved. It is expected that the etch rate of the PECVD silicon nitride layer in BOE is, highly variable depending upon the processing conditions at which this nitride layer was deposited (a common issue in wet etching of PECVD layers) as well as depending upon processing which the device was exposed to after the deposition of the passivation nitride.

Variations in required etch time have even been observed between samples formed from SiC versus Si(111) substrates. The etch time required to clear the surface of the sample is also highly dependant upon the architecture of the device being deprocessed, with submicron "T-gate" devices being observably more difficult to clear of all silicon nitride than devices formed with a gate length of 1µm. The cause of this will be explained in more detail later in the section.
It bears noting that HF is not perfectly selective to silicon nitride over AlGaN. Surface features can develop on the AlGaN surface when a sample is over-etched in BOE. As shown in Figure 2-5, the morphology of these surface features is highly dependant upon the substrate which the the GaN/AlGaN layers were grown on. Epitaxial layers grown on SiC will develop a wavy, "tidepool" morphology if they are significantly overetched. This morphology may follow the terracing of the underlying epitaxial layer. Epitaxial layers grown on Si(111) develop the more traditionally recognized etch pits when overetched in BOE. This morphology results from BOE preferentially attacking the epitaxial material surrounding the threading dislocations which permeate the epitaxial layers of the device. This is a well-documented phenomenon in silicon etching, where BOE has been observed to attack the high-energy surfaces of dislocations [47].

Because of the inherent variability in silicon nitride etch rates, it is difficult to eliminate the surface features which develop as a result of this over-etching. However, they can be minimized if the etch of the device under investigation is minimized. A 15 minute exposure in BOE generally accomplishes the task of etching the passivation layer without attacking the surface dramatically.

With the passivation nitride removed from the surface of the HEMT and with the gate and ohmic contacts exposed as a result, the metal layers can removed to fully expose the surface of the AlGaN epitaxial layer. This is accomplished using a combination ferric cyanide (FeCN) and potassium iodide (KI) etch solution, known as TFAC, which is commercially available from Transcene Company, Incorporated. TFAC is marketed as a gold etchant, specifically designed for use in the liftoff processing of metal layers on compound semiconductors, where an etch is required which can remove deposited metal without attacking the exposed semiconductor layers present underneath the metal. The KI component of this etch solution actively attacks and
dissolves any metal, excepting platinum, while maintaining the quality of the AlGaN layer below. No etch-related defects have been observed in the AlGaN as a result of etching with TFAC, which compares favorably from previous results using Aqua Regia to remove the metal layers.

The FeCN/KI mixture is received after purchase as a dry powder and must be dissolved in water to form a proper etch solution. This is accomplished by mixing a volume of 3.75L of water for every 225g of FeCN/KI and stirring at room temperature using a magnetic stir rod. The FeCN/KI mixture does not dissolve easily in water and stirring may take up to 6hrs in order to fully dissolve the solid TFAC mixture. The rate of dissolution does seem to be improved if the FeCN/KI solid mixture is ground down via a mortar and pestle prior to mixing with water, as dissolution appears to be aided by smaller particle sizes and an associated larger surface area per volume of solid TFAC powder.

Generally, the majority of the contact metal present within the channel of the device can be removed with an etch lasting 24hrs. In order to fully clear the ohmic contact regions of the device, a longer etch (sometimes as long as 96hrs), may be required. An etch which requires a substantially longer time period than this is generally indicative of incomplete removal of the silicon nitride passivation layer. An SiN layer which has not been completely removed will not interfere with the etching of the gate metal (though it may interfere with imaging of the region under the gate once the metal has been removed), but does interfere with the etching of the ohmic metal layers. In such an event, the SiN passivation layer must be re-exposed to BOE and removed. Generally, 5 additional minutes of exposure in BOE will fully remove an SiN layer which was not completely etched by the first 15 minute BOE exposure.
After metal etching is completed, the two layers which obstruct the AlGaN epitaxial layer from direct viewing have been, ideally, removed. All that remains is final cleaning of the surface of the device. This consists of one or two steps, depending upon the device being analyzed. In the case of a micron-scale device, the only additional deprocessing which is required is cleaning in a progression of organic solvents. This exposure is designed to eliminate scumming due to organic contamination on the surface of the sample. It is important to note that, while the methodology described here for eliminating organic scumming is effective, the best deprocessing results consistently come from devices which have been manipulated in a clean environment and with clean lab equipment.

As shown in Figure 2-6, AlGaN surfaces which have been exposed to the BOE and TFAC solutions are, by no means, clean after etching. Organic scumming (seen here as the dark features which are scattered throughout the sample surface, is ubiquitous on most devices. This scumming is enhanced by dirty processing conditions, but it cannot be eliminated even by the cleanest handling procedures. The only solution to this scumming is exposure in organic solvents to loosen and dissolve the scumming films. It should be noted that the effectiveness of this descumming process is greatly enhanced by the use of ultrasonication as part of the cleaning regimen. All solvent exposure occurred in a Fischer Scientific FS60D ultrasonicator.

Organic solvent exposure begins with a 2 hrs exposure in a 1:1 mixture of n-heptane and acetone, which are used to dissolve and break up the bulk of the organic film in preparation for an additional 2 hrs exposure in methanol, which leaves a pristine, but hydrophobic AlGaN surface. An exposure in an equal parts solution of heptane and acetone is preferable to acetone alone, resulting in a much cleaner sample surface. This hydrophobic surface is, unfortunately, incredibly effective at attracting dust present in the processing environment. With this in mind,
the last step in the solvent cleaning process is a 2 hr ultrasonication in water, which further cleans the sample and also generates a hydrophilic AlGaN epitaxial surface, which attracts much less dust from the ambient and which is ideal for imaging with TEM/SPM.

In addition to exposure in organic solvents, submicron devices also require a secondary BOE exposure. In a submicron device, the overhang of the T-gate structure over the conformally deposited silicon nitride allows it to mask this region of the deposited nitride from exposure to BOE. As shown in Figure 2-7, this masking results in the formation of an silicon nitride "stringer", which runs parallel to the gate width of the device. Stringer formation is, generally, not an issue for larger gate lengths and, as a result, no secondary BOE exposure is required for larger gate length devices. Because this "stringer" is so well masked by the submicron gate necessitate a second exposure in BOE after removal of the gate using a metal etch solution. This secondary BOE exposure is timed at 5min for submicron devices which are under inspection. After this secondary BOE exposure, the devices are cleaned in the same progression of organic solvents as was previously described for the case of the 1 µm gate length devices.

This method of chemically deprocessing a device of interest would be of little real use if the top layer of the AlGaN was eroded away as a result of etching. With that in mind, the cross-sectional thickness of the AlGaN layer was analyzed with TEM, using the University of Florida's JEOL 2010F Transmission Electron Microscope in dark field mode. A TEM cross-section was cut from a deprocessed device, which was previously coated with thermally evaporated carbon as a protective layer, and compared to a control sample cut from the gate region of a pristine control device. As shown in Figure 2-8, the thickness of the AlGaN layer does not change as a result of exposure to the HF/TFAC chemistries or from organic solvent exposure. The deprocessing chemistry used is perfectly selective.
2.4 Scanning Electron Microscopy

Scanning Electron Microscopy (SEM) is an electron imaging technique which has been employed in materials studies for a variety of applications since the techniques development in the mid-1930's. The technique is particularly popular in the semiconductor industry, where it is used in both non-destructive analysis of devices at key stages during the manufacturing process (such as dry etching and lithography), as well as in destructive post-mortem analysis of defective devices [48]. SEM actually encompasses a variety of analytical techniques utilized heavily in material analysis. All of these techniques involve scanning an electron beam over the surface of the material of interest and collecting some observable signal which results from the interaction of incident electrons with the material being analyzed. This technique is performed in a vacuum, but the vacuum level employed often varies depending upon the specific system being utilized as well as the application of interest to the user.

An SEM system consists of four modules which act in conjunction to provide an electron image of the material being analyzed [49]. The sample chamber and electron column represent the module where all important interactions between the electron beam and the material under study occur and are measured. However, it would be impossible to induce these interactions, or to measure them, without the vacuum system which is used to pump atmosphere out of the sample chamber and increase the mean free path of electrons to a point where they can be collected by the detectors within the sample chamber. Likewise, the intricate electromagnetic lenses of the electron column would not yield a usable image without the scanning electronics used to raster the beam across the sample surface and to correlate the changing signal from the detectors being used to the changing position of this rastered beam.

Material samples meant for study with SEM generally yield higher quality images when they are conductive and connected to some ground electrode. Without this grounding
connection, sample charging can occur, resulting in distortion of the output image during analysis, which generates an electric field that interferes with the electrons' path to the sample. Grounding can be accomplished by a variety of methods. Carbon tape or silver paste are both commonly used to affix samples to the sample stub placed within the SEM system prior to analysis. These two adhesion methods are desirable because they also form an electrical path from the back surface of the sample to the stub and, through it, to the grounded clamping mechanism located in the sample chamber. This method is effective in the case of an AlGaN/GaN HEMT sample, but the presence of a semi-insulating layer on the near surface of the sample (i.e. the AlGaN layer) still predisposes the sample to charging during analysis. In order to eliminate this issue, carbon tape is affixed to the stub such that it bridges the top surface of the AlGaN and the carbon tape holding the back surface of the sample to the stub. This carbon tape contacts the n-GaN layer present on top of the AlGaN, shorting it and ensuring that charge injected into the sample has some pathway to ground. Utilizing the n-GaN capping layer is desirable over coating with a conducting film, which would impede direct imaging of the semiconductor surface.

Once the sample is properly grounded, it is mounted within the SEM and the system is pumped down to its ideal operating pressure. Depending upon the SEM, this pressure can vary from near-atmosphere to Ultra High Vacuum. In the case of the SEMs used for analysis in this study, this was accomplished by means of a combination of a rough pump and turbo pump, designed to bring the system down to a pressure of $5 \times 10^{-5}$ mmHg.

Once the sample chamber is pumped down to a suitable vacuum, the electron source can be turned on. An aperture opens between the sample chamber and the electron column, which is maintained at a pressure lower even than that of the sample chamber. Electrons with an energy
defined by the instrument's operator are emitted from a source at the top of the column and are accelerated through a potential difference before encountering a series of electromagnetic lenses designed to condense the image of the electron source into a beam of uniform illumination and demagnify this beam as much as possible. It is in this region that the steering electrodes can be found, which raster the electron beam back and forth over the sample surface. The overall beam current passing through this column is controlled by an aperture placed between the end of the electron source and the beginning of the lens optics of the electron column.

It bears noting that the lens optics located in the electron column must, generally, be tuned by the user to induce an image with the maximum resolution. The incident beam may be adjusted to yield a perfectly circular spot, as opposed to an ellipsoid, by "wobbling" the electron beam optics in and out of focus and varying the x and y stigmation until the image produced defocuses evenly in both directions. The alignment of the electromagnetic lenses present within the electron column is also critical for right resolution imaging. This alignment is accomplished, again, by wobbling the beam optics in and out of focus and adjusting the lens alignment in x and y such that the resultant image does not shift while the beam is being "wobbled".

A variety of electron sources have been utilized in SEMs since their inception. The most popular of these sources utilized in state of the art systems is the thermal field emission source, which is capable of producing an electron beam which is smaller, more coherent and energetically uniform, and with three orders of magnitude higher current density than older sources such as sources which use thermionic emission. Modern SEMs are capable of a wide range of accelerating voltages and generally operate in the 500V to 30kV regime and are capable of demagnifying the source to an electron beam which approaches 1nm spot sizes. The SEMs used in this study utilized a thermal field emission source and the extracted electron beam was
accelerated through a voltage of 5kV. In each system, an aperture was selected which
maximized current without hindering high resolution imaging. The FEI NovaSEM, utilized for
some imaging as part of this study, operated in a regime capable of generating a 1.0 - 1.5nm spot
size electron beam, while the FEI DB235 FIB, which was utilized for most of the imaging
performed in this work, operated in a regime capable of generating approximately a 5nm spot
size electron beam.

The mean free path of the electrons which leave the column is large enough that they do
not interact with any other matter en route to the sample. When they encounter the material
under study, the electrons are slowed by interactions with the nuclei and electron clouds of the
atoms which make up the sample and their kinetic energy is deposited into the material itself as
the electrons are decelerated. The result is a characteristic "teardrop" which represents the
interaction volume that a given injected electron has with a material sample as it follows a
brownian path of travel influenced by interactions with the sample atoms.

Different resultant information, in the form of three separate characteristic electron
signals and two characteristic light signals, are emitted from different regions of the interaction
volume depending upon the specifics of the interaction which occurs. Of these, only three
separate signals (backscattered electrons, secondary electrons and cathodoluminescent light) are
germane to data represented in this work. The reader is directed to other texts if signals other
than the three listed above (characteristic and continuum x-rays and auger electrons) are of
interest [50]. The size of the teardrop, itself, increases with increasing incident electron energy
as well as with decreasing atomic number. As these two criteria are met and the interaction
volume grows in size, it does so in both the vertical direction (penetration depth) and in the
lateral direction (spatial resolution). It is for this reason that all imaging performed on the
materials studied as part of this work was performed at a moderate accelerating voltage (5kV) rather than at larger accelerating voltages.

As electrons pass through the sample material, they encounter and interact with the atoms which comprise that material. When an electron encounters an atom, it can glance off of the electron cloud of that atom, imparting some of its kinetic energy in an inelastic collision and slowing down as a result. The other option available to the electron is to interact with the bulk of the electron cloud. If this occurs, the incident electron recoils violently in an inelastic collision which ejects it from the sample material. These electrons are called backscattered electrons and, not surprisingly, they are highly sensitive to the number of electrons present in the cloud of each atom. As atomic number goes up and the density of electrons goes up with it, the yield of backscattered electrons increases. Because they are the result of the collision of a material interaction with an incident electron which lost little energy, backscattered electrons can be found throughout most of the interaction volume.

In the case of an inelastic scattering event, the interaction between the incident electron and the electron cloud of the atoms comprising the sample can result in an electron being kicked free of its parent atom with a very low energy (<50eV). Secondary electrons located near the sample surface can be emitted from the material with high probability, yielding substantial information about the topography of the sample. If a secondary electron isn't emitted, it can be captured by an atom and fill an unoccupied orbital. In semiconducting and semi-insulating materials, this process occurs with high frequency, as a large number of unoccupied orbitals exist in the form of holes within the material. This electron-hole recombination event is known as cathodoluminescence [51]. Secondary electrons are generated close to the surface of the sample (within 5nm to 50nm), where the electrons incident to the sample surface have not yet
lost substantial kinetic energy and may still undergo a large number of inelastic scattering events. The position of the secondary electrons in the thinner, necked region of the teardrop near the surface results in a small lateral straggle for the signal, enhancing the spatial resolution of the measurement being made. For this reason, all imaging performed in this study was performed using detector settings meant to enhance the collection of secondary over backscattered electrons.

A standard secondary electron detector (SED) sits off axis from the electron column in order to detect the electron signal from the sample without interfering considerably with the demagnified image of the source as the incident electron beam approaches the sample from the source. The SED consists of a phosphor-based scintillator covered with a faraday grid and in line with a light guide and photomultiplier tube. As secondary electrons are ejected from the sample surface, they are pulled towards the secondary electron detector by the faraday grid, which is held at a positive bias of, at maximum, a few hundred volts. These electrons are bent towards and accelerated through the faraday grid and impact the scintillator. The scintillator fluoresces and the escaping light is channeled down a light guide towards the photomultiplier tube, which amplifies the signal from the scintillator so that it is more readily detected by the photodiode mounted at the end of the tube [52].

This is an effective method for detecting the secondary electron signal emanating from the sample material. However, imaging a secondary electron signal at high magnifications is far more challenging because the probe size, and resultant number of secondary electrons output from the sample, is small. To that end, a different method of collection is required for high resolution SEM with secondary electrons. The solution is the use of a through-lens detector (TLD), which is a scintillating electrode mounted directly to the pole piece of the microscope,
where the electron beam generated in the source and manipulated by the column enters the sample chamber. This positively-biased electrode sits much closer to the sample than a conventional SED, and is more readily able to detect a secondary electron signal [20]. In modern systems, such as the FEI Nova SEM utilized in this study, the TLD is coupled with an immersion lens, so called because it physically envelops the sample surface in an electromagnetic field which induces a helical path of travel for secondary electrons leaving the sample, increasing the efficiency with which the TLD located within the pole piece of the microscope collects secondary electrons and screening out the backscattered electron signal. This immersion lens also screens out ambient magnetic fields present within the sample chamber of the microscope and the environment in the area where the microscope itself is installed, eliminating their deleterious effects on imaging. For this reason all imaging performed on samples studied with SEM is performed using the TLD, when available.

2.5 Surface Probe Microscopy

Surface Probe Microscopy (SPM), also known in the literature as Atomic Force Microscopy (AFM), utilizes a nanoscale probe located at the tip of a micromachined cantilever beam to measure the properties of surfaces on scales ranging from the micrometer to the nanometer scale [53]. An SPM system consists of three major modules which work together to form an image of a materials surface [54].

The first of these modules is the "head" of the SPM system, which is an assembly into which the SPM cantilever which is used for imaging is attached. The tip itself is moved about the sample surface using piezoelectric actuators located in the SPM head which are capable of shifting the cantilever in the lateral and horizontal directions. Another piezoelectric actuator is used to control the vertical deflection of the cantilever as it travels over the sample surface. This deflection is measured by bouncing the beam emitted by a laser diode off of the tip of the
cantilever. This bounced beam strikes a four-cell photodiode and the resulting difference in induced photovoltage between each cell is used to measure the total deflection of the cantilever beam during its travel.

The deflection measured by means of the four-cell photodiode is fed into the second major module of the SPM system, which are the computer-controlled electronics used to control the cantilever tip and to collect deflection data resulting from the measurement being taken. In addition to a Personal Computer, which acts as the operator interface for the AFM, a lock-in amplifier is used to measure the signal output from the four-cell photodiode and a frequency generator is used for input signals to the piezoelectric height actuator used to induce probe deflection. The function of the control electronics varies depending upon the mode of imaging selected for the experiment.

There are a variety of methods which can be used to image a surface with SPM, but they all fall, more or less, into two separate imaging modes. In contact mode, the probe tip of the SPM cantilever is put in direct contact with the sample surface, such that a repulsive force is generated between the sample and the probe, resulting in cantilever deflection. The control electronics work to maintain this deflection, and the magnitude of the repulsive force which generates the deflection, at a constant value. In tapping mode, the SPM cantilever is vibrated above the sample surface at a frequency just higher or just lower than its resonant frequency. As the tip approaches the sample surface, interactions between the probe at the end of the SPM cantilever and the sample surface induce a change in the resonant frequency of the AFM tip, which is detected by the control electronics. In tapping mode, the control electronics maintain the deflection of the AFM tip in such a state that the resonant frequency of vibration does not change as the tip is scanned over the material of interest. This changing deflection can be
converted into a map of the interactions at the sample surface. In the case of a surface which does not yield and which does not cause the SPM tip to "stick", as is the case for the AlGaN surface of a HEMT, this map of interactions can be assumed to be equivalent to the varying height of the surface of the sample relative to the scanning SPM tip.

The third, and arguably the most important, module of the SPM system is the SPM cantilever itself, which is generally a strip of thick silicon or silicon nitride material several millimeters long by several millimeters wide. A "T" structure is typically etched out of the top surface of this cantilever to mark the desired position of the structure for mechanical clamping to the SPM head while the probe is located at the tip of the cantilever on its "bottom" side. Sometimes, the top surface of this cantilever will be coated with a reflective thin film in order to maximize the reflected optical signal travelling from the cantilever to the photodiode-based detector of the SPM.

The material which this tip is formed from and its general dimensions can vary dramatically based on the application desired. Nonconductive probes with a high aspect ratio and a small radius of curvature are ideal for the measurements of nanometer-scale features on smooth surfaces, as is the case for defects present within the AlGaN/GaN material system. These tips are generally formed from the cantilever itself, which can be reactive ion etched to form an ideal tip geometry, or from the direct bonding of nanotubes or nanowires to the underside of the cantilever beam.

A variety of imaging artifacts can influence the measurement of surfaces using SPM and lead to spurious interpretation of data. Some of these artifacts can be eliminated by properly adjusting the various settings of the control electronics, chiefly the proportional and integral gains, so as to ensure the proper travel of the SPM tip across the sample surface as associated
recording of the properties of that surface. Generally, the overall quality of the image formed by SPM analysis can also be improved if the scan rate of the SPM tip across the sample surface is reduced, allowing the cantilever more time to equilibrate and yield a low-noise signal for a given pixel before transitioning to the next point in the image.

The SPM tip used for analysis can also have a large bearing on the quality of the image formed as the tip is rastered across the sample surface. The two chief metrics which have a bearing on overall image quality for a given SPM tip are the aspect ratio of the tip itself and the nominal tip diameter. SPM cantilevers which have large probe tip diameters are prone to imaging artifacts because a large tip area interacts with the surface at any given point in time. Not surprisingly, this effectively limits the resolution of a given micrograph to the tip diameter and results in small surface features being represented as having larger dimensions than is physically the case. Small features which are inset from the sample surface, such as pits or trenches, may be rendered effectively "invisible" in this case, as the SPM cantilever tip may travel directly over the recessed feature without detecting it due to a large tip diameter. Tips with a low aspect ratio can also lead to imaging artifacts when imaging densely-packed features possessing a high aspect ratio. In this case, the tip may interact with the sidewalls of a recessed or feature rather than its bottom surface, resulting in questionable topographic data. As might be surmised from the discussion above, it is of critical importance to utilize an SPM cantilever tip which possesses a nominal tip diameter and aspect ratio which matches the dimensions of the features under observation.

All SPM imaging performed as part of this study occurred on a Bruker Dimension 3100 Atomic Force Microscope in tapping mode, with a deflection voltage ranging from approximately 300mV to 400mV, depending upon the sample under observation, and with the
integral gain and proportional gain for the feedback electronics set at values of 5.0. Prior to measurement, the system was calibrated so that the observed difference in four-cell photovoltage was below 100mV. The scan rate for image capture was set at 0.5Hz and 512x512 pixel arrays were generated, corresponding to a height map of the AlGaN surface. A Bruker TESP-HAR tip with an 5:1 aspect ratio, 40 N/m stiffness, and 10nm nominal tip diameter, was utilized for all measurements.
Figure 2-1 A schematic view of the electron optics utilized in a typical TEM. Visible are the electron source and condensing optics, used in forming a coherent source for imaging, as well as the objective and intermediate optics, which determine the information about the sample which is transmitted to the projector lenses and screen.
Figure 2-2 The steps associated with Focused Ion Beam milling of a generic sample. Progression runs down the left column and then down the right column. The configuration of the sample relative to the ion and electron beams on the left and a top down image of the sample is on the right. After deposition of a protective mask via an ionized organomettallic beam and trench milling to either side of the sample, the sample is tilted relative to the gallium beam in order to perform an undercut and release one side of the lamella. An omniprobe needle is inserted, and welded to the sample. After welding, the last side of the lamella is cut free and the sample is retracted along with the omniprobe needle.
Figure 2-3  A top-down image of a typical maskset analyzed as part of this work. Three separate device architectures are utilized for combined electrical and microscopy analysis. These are the "FatFET" (a large device with a 75µm gatelength), a submicron device (with a gatelength ranging from 100nm to 160nm) and a micron device (with a gatelength of 1µm).

Figure 2-4  The Deprocessing of an AlGaN/GaN HEMT. Top-down SEM micrographs with corresponding schematic cross-sectional representations (not to scale) show the etching strategy employed for deprocessing AlGaN/GaN HEMTs to allow quantitative under-gate defect analysis prior to deprocessing (left), etched for 15 min with BOE (middle), and subsequently etched for 18 h with TFAC, etched for 5 min in BOE for a second time, and cleaned ultrasonically for 2hrs in a volumetrically equal mixture of acetone and n-heptane, for 2hrs in methanol and 2hrs in water (right). The locations of the source (S), gate (G), and drain (D) contacts for each image are indicated.
Figure 2-5 AlGaN surfaces after exposure to HF. A) Plan-view SEM images of a sample etched with BOE until full removal of SiN with minimal over-etching. B) Plan-view SEM of a sample formed from a SiC seed crystal which was over-etched and to reveal a "tidepool" morphology. C) Plan-view SEM of a sample formed from a Si seed crystal which was over etched and which formed more characteristic etch pits.
Figure 2-6 AlGaN surfaces after exposure to various organic solvents.  
A) Plan-view SEM images of a sample as-deprocessed, with no further exposure to organic solvents.  
B) Plan-view SEM of a sample exposed for 2hrs in methanol  
C) Plan-view SEM of a sample exposed for 2hrs in 1:1 acetone and n-heptane and for 2 hrs in methanol.
Figure 2-7  Incomplete Removal of SiN from a T-gate. This is due to masking by the metal gate, resulting in the formation of a "stringer" which must be removed with a secondary etch in BOE after TFAC exposure.

Figure 2-8  HAADF-STEM images of deprocessing of the near-gate region of a HEMT. A) The HEMT structure is shown prior to deprocessing, with Au and Ni metal layers. B) After deprocessing, thorough removal of the metalization layers with the AlGaN and GaN epilayers left completely intact. The protective C layer was deposited prior to sample preparation using focused ion beam milling to prevent surface damage.
Figure 2-9  A schematic diagram of a basic SEM system. The sample chamber, control and display electronics, and electron column are all present. The vacuum system (not pictured) maintains pressure in the electron column as well as in the sample chamber.

Figure 2-10  A schematic diagram of a generalized SPM system. The SPM head produces a deflection signal which is controlled by an set of negative feedback loops, adjusted via a user-controlled PC. In this schematic, the tip is contained within the SPM head, though it is a modular component.
Ohmic contacting schemes to compound semiconductors are quite varied in their nature, ranging from traditional methods including doped wells, to a variety of alloyed contacting schemes, and even to nanostructured contacts which utilize thin interfacial layers to alter the energetic barrier to charge transfer. It is not surprising, therefore, that different contacting styles utilized for making an ohmic connection to different semiconductors should suffer from different degradation mechanisms.

Figure 3-1 demonstrates the contacting scheme utilized in the AlGaN/GaN HEMT structures studied as part of this work. The ohmic contacts utilized in this work are separated from each other by approximately 4.0µm and from the gate electrode of the device by a maximum of 2µm and sometimes as little as 1.5µm, depending upon the geometry of the gate contact of the device. The contacts, themselves, are formed from a traditional metal stack utilized in contacting: Ti/Al/Ni/Au. These metals mix together to form a metallic alloy during Rapid Thermal Annealing (RTA) at 850°C for 30s. The result is the formation of an ohmic contact with a contact resistance ranging from approximately .2Ω mm to 1.5 Ω mm [19].

This RTA process is known to greatly impact the overall quality of the ohmic contact made to the AlGaN surface and has been linked to changes in the surface roughness of the contacting metal pad [55]. Deprocessing of ohmic contacts via metal etching has also demonstrated that this contacting scheme appears to rely on the diffusion of some component of the ohmic metal into the underlying AlGaN epitaxial layer, presumably down the threading dislocations which extend up from the growth substrate, through the MOCVD grown GaN and into the AlGaN [6]. This diffusion event results in the formation of a metal inclusion, which
shorts out the ohmic contact, resting on top of the AlGaN epitaxial layer, to the 2DEG which is extant at the interface between the AlGaN and GaN.

The first section of this chapter will focus on previous studies into the annealing of various metal stacks formed on AlGaN/GaN in order to form an ohmic contact to the buried 2DEG layer. Special attention will be paid to intermetalllic reactions which both aid in conduction between the 2DEG and the ohmic contacts and which hinder this conduction through the formation of resistive phases and the degradation of contact morphology. The second section of this chapter will focus on the structural nature of this metal inclusion and will discuss the conditions which it induces the surrounding epitaxial materials during its formation. These conditions, specifically the large compressive stresses induced in the epitaxial layer as the result of the formation of this metal inclusion as well as the tensile stresses induced at the edge of these inclusions, influence defects formed during processing which hinder device performance. The second section of the chapter will focus on the defects which form during processing. These defects are nanocracks which originate at the ohmic contact and which can penetrate into the channel material of the HEMT, adversely effecting charge transfer characteristics within the device. Evidence also exists which suggests that these defects can be influenced by electrostatic stressing of the device.

Devices under analysis as part of this study were formed using the same processing conditions and device architectures specified in Chapter 1. AlGaN/GaN HEMTs with a 100nm gate length were used. In order to observe the formation of these defects, two separate analytical tools were employed. Lamellae for cross-sectional TEM analysis were formed via FIB/SEM on a FEI DB235, where in-situ sample milling, liftout, mounting, and thinning were accomplished via the use of an Omniprobe needle. The ion source used for milling was a gallium beam. All
samples were coated with a 200 nm carbon layer, deposited via thermal evaporation and an organometallic Pt source was used to form the 2µm protective mask utilized during milling. Analysis of lamellae formed via FIB/SEM was performed on a JEOL2010F TEM, with an EDS attachment. Analysis of the sample was done using HAADF-STEM because of its utility in quantitative analysis using EDS. Samples under analysis were also studied using top-down SEM after the deprocessing methodology described in Chapter 2 was performed. Imaging of deprocessed samples was performed on the same FEI DB235 in Ultra High Resolution mode with the Through Lens Detector selected.

As mentioned previously, all stressing and measurement of device characteristics was accomplished via the use of an HP4146C device tester, capable of independent measurements of current and voltage on all three device contacts, and a Techtronix 370A curve tracer. The devices used in this study were formed using the processing described in the first chapter of this dissertation on a SiC substrate. Stepped stressing was performed on one device analyzed as part of this study, where the device was taken from a value of $V_{GS} = -10$V to $V_{GS} = -42$V at -1V/min with $V_{DS} = 15$V. Measurements were made after every voltage step with a duration of 1ms, making the duty cycle of the measurement effectively equal to 100%.

### 3.1 Progress in Ohmic Contact Annealing

In addition to their characteristic low thermal conductivity [57] and semi-insulating behavior [1], GaN-based systems are highly unreactive [58]. This stems from the highly ionic nature of bonding in GaN and AlGaN, which contributes to all of the qualities listed above. Because of the structure of AlGaN/GaN HEMTs, where the conductive channel is buried under a thin layer of semi-insulating AlGaN, the unreactive nature of this material system makes ohmic contacting a challenge [59].
Current contacting schemes to AlGaN/GaN HEMTs rely on a multi-layer stack of deposited metals which are annealed in order to induce alloying within the contact and between the contact and the AlGaN/GaN epitaxial layers. This metal stack initially was formed from a layer of titanium placed in contact on the AlGaN surface, followed by a layer of aluminum, both deposited by electron beam evaporation or sputtering [60]. Upon annealing at temperatures ranging from 750ºC to 950ºC, the titanium undergoes a reaction with the Al layer above it as well as the AlGaN below it, forming TiN as well as AlTi$_2$N, which are formed via a reaction where nitrogen is depleted from the AlGaN epitaxial layer. This reaction results in the formation of an electron-rich surface layer in the AlGaN which results in enough band bending within the AlGaN to form an energetic barrier which permits the tunneling of electrons [61]. The TiN formed as part of the annealing process also results in enhanced conduction. In some cases, it expands into the AlGaN layer down threading screw dislocations to form an inclusion [61]. These inclusions reduce the distance between the ohmic contact and the 2DEG present at the interface between the AlGaN and the GaN and reduce the resulting tunneling distance for electrons. In some cases the metal inclusion completely bridges the distance and forms a direct link between the two charge-conducting layers.

This alloying scheme is not without its flaws, however. Titanium and Aluminum are highly reactive metals and oxidation results from contact annealing, even during rapid thermal annealing in a nitrogen-purged environment [62]. Contact degradation due to oxidation can be averted by coating the ohmic contacts with a noble metal. Gold is the contacting metal of choice for this application. However, gold intermixes with the Aluminum and Titanium in the ohmic contact during annealing, resulting in the formation of a variety of intermetallic phases, observed via EDS and STEM, including Al$_2$Au [63], AlAu$_4$, AlAu, and AlAu$_2$Ti [64]. None of these
phases appear aid in conduction out of the ohmic contact and their formation often degrades the morphology of the thin film stack, resulting in substantial roughening which also degrades the contact resistance of the device.

In order to avert the formation of these intermetallic phases, which greatly degrade the overall quality of the device, a barrier layer is often inserted into the device structure. This barrier layer is used to act as a means of preventing Au diffusion into the Al and Ti layers of the ohmic contact. A variety of metals have been utilized for this application, including Pd [65], Ti [66], Mo [67], and the barrier layer used in the devices studied as part of this work: Ni [68]. Needless to say, the metals used in barrier layers have variable efficacy in addressing the problem of Au diffusion and reaction with the underlying Al and Ni layers and in doing so without suffering from some other adverse reaction which degrades the character of the electrical contact.

Ni, which has been the barrier metal of choice up until recently, is a perfect case of a barrier material which enables contacts with very low resistivities (approaching 0.25 Ω-mm) but which suffers from a deleterious secondary reaction which greatly impacts the surface morphology of the contact, making soldering during later packaging processes difficult [69]. multiple XTEM studies have show that, upon annealing, Ni reacts with Al and, in doing so, ceases to behave as a contiguous layer, allowing Au to diffuse down into the underlying Al to form intermetallic Au-Al phases. The film stack roughens substantially upon annealing as the Ni-Al and Au-Al phases separate from one another [70]. Mo, which is becoming the barrier metal of choice for contacting, does not react with Al and does not disintegrate. However, Au is capable of diffusing through the grain boundaries of the metal thin film to interact with the Al below. The action of this diffusion degrades this barrier layer at high temperatures (950°C).
A variety of methods exist to improve the efficacy of these layers, through optimization of different metal layer thicknesses, to optimization of annealing conditions and even to addition of additional materials to form a more complex alloyed contact. It stands to reason that, all other things being equal, a thicker barrier metal layer should reduce the amount of Au which intermixes with Al. Ideally, a barrier layer should be thicker than the diffusion length of gold within that layer. Studies performed on a Ti/Al/Ni/Au stack demonstrated that the contact resistance and observed line edge roughness of contacts were reduced for Ni layers which were two times as thick as the deposited Ti, Al and Au contacting layers, with a contact resistance of 0.26 Ω-mm. The morphology of the contact was found to be dominated largely by the thickness of Au. Increasing Au thicknesses resulted in ever growing surface roughnesses, presumably due to Au balling. Contact resistance goes through a local minimum at a Ni thickness of 1.8 times the Ti and Al thicknesses, likely because Ni acts as an efficient diffusion barrier to Au without consuming substantial Al volume [71].

Annealing of the ohmic contact is critical to the formation of TiN inclusions which short the AlGaN by joining, via tunneling or direct metallurgical connection, the 2DEG to the alloyed contact. Alloyed contacts formed from Ti/Al based metallurgy go through a local minimum of resistance, as measured by TLM pads, in the range of 800°C to 850°C. A sharp rise in contact resistance occurs at temperatures above this range. This local minimum in contact resistance is generally believed to be induced by the reaction of Ti and Al with the AlGaN or GaN, while the sharp rise in resistance which occurs above 850°C is due to intermetallic reactions between gold and the rest of the alloyed contact. A multi-step annealing process performed on a Ti/Al/Ni/Au metal stack, consisting of three 45s RTA cycles at 400°C followed by 700°C and finished at 830°C, has been shown to greatly improve the surface morphology of alloyed contacts as well as
their contact resistance. This improvement has been attributed to enhanced reactivity of Al with Ti and GaN at lower annealing temperatures, which reduces the Al available for reaction with Au [72].

Recent research has also shown that the addition of a sputtered Si layer to the alloyed metal stack seems to enhance conductivity of the ohmic contacts such that the contact resistance of the alloy dropped from .3 Ω-mm to .2 Ω-mm after annealing at 750ºC. Analysis with XTEM demonstrated that this reaction was likely caused by intermetallic reactions between the silicon and the rest of the alloyed contact which resulted in a reduced reaction between Al and Au [73].

3.2 Metal Inclusions and Nanocrack Formation

Figure 3-2 is a representative image of a metal inclusion formed in a HEMT as a result of the ohmic contact annealing process which was previously described. The top of the HAADF-STEM image is representative of the metal pad utilized in forming the source-drain ohmic contacts. Visible within this metal layer are columnar grains of TiN, which phase separates from the other components of the ohmic contacts upon annealing. Below the deposited metal layer and these TiN grains is the surface of the AlGaN and, below the AlGaN, the MOCVD grown layer of GaN upon which the AlGaN sits. A metal inclusion is visible in the image, extending from the AlGaN surface down into the underlying GaN. A threading dislocation is visible extending from the bottom surface of this metal inclusion into the Ga.

The presence of this threading dislocation under the metal inclusion is consistent with previous reports regarding the annealing of metal contacts on AlGaN/GaN epitaxial structures. Metal inclusions form by the diffusion of metal atoms down high-energy threading dislocations, where the activation energy for diffusion is reduced. The AlGaN and GaN are not capable of supporting the high metal concentrations within these threading dislocations as metal diffusion
proceeds, resulting in the formation of an inclusion rather than a metal-rich "pipe" of AlGaN or GaN.

It is worth noting that several distinct layers of material are observable within this metal inclusion, as was the case in previous research on Ti-based ohmic contacts. Special attention should be paid to the signals arising from the Ti Kα, Al Kα and Ga Kα peaks in the EDS linescan of the metal inclusion. These signals indicate that the inclusion is comprised largely of titanium – likely in the form of TiN, given the observations made by previous researchers. Gallium is conspicuously absent from this inclusion and has likely been displaced. An aluminum layer, either formed from the aluminum in the metal contact itself or (more likely) by the aluminum which made up the AlGaN layer displaced by the inclusion itself, surrounds the TiN inclusion.

Figure 3-3 demonstrates the result of the formation of these metal inclusions. After deprocessing, top-down SEM revealed three distinct topographical regions on each of the mesas upon which the transistors under analysis were formed. Two of these are regions where ohmic contacts were formed: the Source and the Drain. Fully 10% of the total aerial density of the source and drain regions is comprised of areas which were once metal inclusions. The average size of these inclusions is 62 nm in diameter. The source and drain are separated by a smooth AlGaN surface which once comprised the channel of the device. These inclusions are hexagonally faceted, not rounded – which would reduce the surface to volume ratio and the free energy in the case of a material with equal surface energies for all orientations – and appear to be bisected by nanocracks observed in the deprocessed contact regions. A larger volume inclusion, where the prism planes of the AlGaN are eroded preferentially is a lower-energy feature than a cylindrical inclusion where all planes of the AlGaN are eroded with equal certainty. This fits well with the substantial bonding energy inherent to the wurtzite GaN.
system, where surface energies vary by many eV per angstrom for different crystal orientations thanks to the highly polar nature of their bonding [74].

A consequence of the formation of these inclusions are nanoscaled cracks which appear to radiate from the corners of the TiN metal into the surrounding volume of the AlGaN. The sharp faceted corners of these metal inclusions appear to be very effective in inducing the initial formation of a surface defect. The average length of one of these nanocracks is 52nm.

If an inclusion exists at the edge of the source or drain, the crack will extend into the channel. In general, such a crack does not extend into the channel much farther than the cracks present within the ohmic contact regions. Cracks present in the channel region are, on average, no more than a few hundred nanometers longer than cracks present within the ohmic contact regions, with an average length of 273nm.

### 3.3 Nanocrack Morphology

Cross-sectional TEM was performed on a nanocrack which was observed in UHR SEM. This was accomplished by, first, marking off the crack with fiduciaries which allowed for proper orientation of the ion beam and associated lamella formed via FIB after the surface of the sample was coated in carbon. This allowed for a cross-section to be formed which runs perpendicular to the direction in which the crack was propagating. An image of this nanocrack and the corresponding orientation of the nanocrack is shown in Figure 3-4. The resulting cross-section of the extended nanocrack was imaged using HAADF-STEM as well as conventional electron diffraction utilizing a parallel beam.

The bulk of this nanocrack is present in the AlGaN epitaxial layer and the total crack depth can be estimated to be approximately equal to 20 nm. It does not extend deeply into the GaN. Thus, the features observed in SEM appear to be channel cracks forming, and more or less constrained, to the thin film of epitaxial AlGaN present on top of the GaN. Electron diffraction
taken from the AlGaN crystal region of the TEM cross-section demonstrates that the crack itself propagates along the [11-20] prism planes of the lattice, which fits well with cracks being present both parallel and at 60° angles to the channel direction, as the channels of these transiting devices are oriented normal to the prism planes of the hexagonal lattice. Crack propagation has been observed along these prism planes by previous studies as a result of film growth well beyond the pseudomorphic limit [75].

The variable length of nanocracks inside and outside of the ohmic contacts can be readily explained by considering the nature of the cracking which is occurring in the epitaxial layer under the channel and under the ohmic contacts. The assumption that AlGaN and GaN share a very similar Young's Modulus and Poisson's Ratio has been made by researchers in the past, when analyzing channeling cracks [76] and the assumption that the fracture toughness of bulk AlGaN is similar to that of GaN also seems reasonable, given this knowledge.

The AlGaN layer in the channel region can be considered to be stress free with the exception of the residual stresses induced in the layer as a result of epitaxy. A crack which forms at the edge of an ohmic contact, in the channel region, will respond to this residual stress and will propagate until the length of the crack is such that the residual tensile stress cannot support additional crack formation at the tip. This effect can be modeled by a modified form of Griffith's Equation [77], where $K$ is the fracture toughness of the film (in Pa·m$^{1/2}$), $c$ is the crack length (in m), and where the tensile stress acting on the film required for additional crack growth is $\sigma$ (in m). It is this variable which can be used to relate back some information about the stress state of the AlGaN under the channel, provided that the geometry of the epitaxial film is taken into account. In this case, a modifier ($Z$) is used to account for the geometry of a crack forming in an epitaxial film on top of a much thicker substrate. This constant approaches a value of $1.976$
for a 2D channeling crack as the film becomes thinner, becoming nearly indistinguishable from this value roughly after a ratio of 1:10 is achieved between the film thickness and the thickness of the relaxed layer below it, which is certainly the approximate case for the AlGaN epitaxial layer on top of GaN [78].

\[ \sigma = \frac{K}{(Z\pi c)^{1/2}} \]  

(3-1)

It is noted that the scenario detailed above is not exactly the physical case in the device under stress, which is coated with a passivating nitride film. However, because the constant resulting from any channeling crack in a thin film varies only between 1 and 2, this approximation is assumed accurate to a first order. The author suspects that the solution for the channel cracking of a thin epitaxial layer with a passivating thin film layer (almost certainly under some unknown plane stress and with an unknown compliance at the nanoscale), is a task meant for simulation. However, the accuracy of this result would be highly dependant on a variety of materials factors which almost certainly vary substantially for a PECVD SiN process [79].

In order to calculate the stress present in the film from a given crack length, the fracture toughness must be estimated. This is accomplished utilizing the following equation, which calculates the fracture toughness in the plane strain state. In this equation, \( \gamma \) is the surface energy of the crack (equal to twice the surface energy of a [11-20] prism plane, or 157 eV/Å\(^2\), assumed by linear interpolation between the theoretically calculated surface energies of the GaN [11-20] and the AlN [11-20] planes, for an estimate) [74,80]. The Young’s Modulus (E) is assumed to be equal to 309GPa [81] and the The Poisson Ratio (v) is assumed to be equal to 0.51 [76]

\[ K = \left(\frac{2E\gamma}{(1 - v^2)}\right)^{1/2} \]  

(3-2)
The average stress calculated from these coupled equations for an average crack present in the channel is 584 MPa, which is roughly a fourth of the value calculated from the difference in lattice constant between relaxed and pseudomorphic Al$_{0.28}$Ga$_{0.72}$N (1.94 GPa), and not a very good fit [82]. Because the crack has been observed in the AlGaN layer away from the metal inclusions, this value of tensile stress can be assumed to be just larger than the average stress encountered by cracks which move through the channel, however, this assumption is made without any knowledge about the distribution of crack lengths inside and outside the ohmic contact regions.

The histogram of crack lengths and corresponding tensile stresses for additional growth associated with these lengths is shown in Figure 3-4. This histogram is representative of approximately 600 cracks present under the ohmic contact and channel regions, each. It originated from 20 devices recieved as-fabricated.

Cracks present within an ohmic contact occupy a very narrow distribution centered around a maximum of 40-60 nm, which is approximately equal to the calculated average. Nevertheless, the histogram demonstrates that this distribution is not normal, but skewed towards larger crack lengths. This skew intensifies dramatically in the case of cracks observed in the channel region, where the maximum is located around 140 nm, which is much smaller than the calculated average of 273 nm. This leads to an estimate of 780 MPa for the stress in the AlGaN film, which is about a forty percent of the tensile stress in the film produced with theoretical calculations – a better fit in comparison to the value derived from the average value of the crack lengths in the channel.

Using the same equation, the stress in the ohmic contact region can be estimated as equal to 1.87 GPa, which is significantly larger. This indicates that the epitaxial AlGaN is under
compression as a result of the formation of metal inclusions. It is this compressive axial stress which first causes the formation of the flaws from which the cracks observed in the channel grow. This flaw formation is due to a tensile hoop stress, oriented radially with respect to the inclusion, induced by the force of this inclusion pushing on the AlGaN epitaxial layer. The compressive stress which results in this flaw formation can be calculated using Equation 3-3 [83], where R is the approximate radius of the metal inclusion itself and set equal to 26 nm.

\[ \sigma = \frac{K}{HR^{1/2}} \]  

(3-3)

In this case, H is a constant associated with the comparative dimensions of the crack initiated by the inclusion and the size of the inclusion, itself, and never exceeds a value of 0.24, provided that the crack's size exceeds half the radius of the inclusion. This assumption is accurate for the case of cracking in the ohmic contact regions. So, the compressive stress induced in the AlGaN by the metal inclusions, and acting on the inclusions to generate tensile hoop stress, may be calculated in this fashion. This results in an estimation of the compressive stress present in the ohmic contact region due to the formation of a metal inclusion as being equal to 38.1 GPa, which is a substantial compressive stress and actually about 23% higher than the compressive stress derived from the strain based purely on the aerial density of pits in the ohmic contact regions (30.8 GPa). The stresses present in the epitaxial AlGaN in the ohmic contacting region can be expected to be larger than this value, as the effects of neighboring inclusions was not taken into account as part of this equation.

It bears noting that only the most infinitesimal amount of additional tensile strain is required within the channel in order to induce additional growth of one of these cracks. It seems plausible, therefore, that electrostatic stressing could induce additional crack growth by means of biaxial tensile strain induced by the inverse piezoelectric effect, which is described in more detail
in Chapter 4. If this occurs, it likely occurs as particularly long cracks approach the gate contact, where the electric field is substantial. This must occur infrequently, as lengthened cracks have only been observed once during the course of the electrostatic stressing studies associated with this work.

Figure 3-6A and Figure 3-6B represent the $V_{DS}$-$I_{DS}$ family of curves for the transistor as well as the $I_{GS}$-$V_{GS}$ relationship for the shottky contact which comprises the gate. They are presented both before and after stressing for a device which underwent stepped stressing from $V_{GS} = -10V$ to $V_{GS} = -42V$ at -1V/min and with $V_{DS} = 15V$. An increase in gate current and a decrease in saturation current can be observed. This could be due to shorting of the gate contact to the source and drain contacts, possibly through the 2DEG.

The nanocracks present within this stressed device were analyzed after deprocessing, as evidenced by the histogram in Figure 3-6C. It should be noted that the implicit assumption is made throughout this text that the formation of nanocracks occurs in the device prior to deprocessing. This may not be the case, but the assumption is made regardless because nanocracks cannot be observed on these devices without the aid of deprocessing. As is shown on this histogram, many of the crack lengths in this stressed device were much larger than crack lengths observed in other devices which were recieved as-fabricated. Some of these cracks have grown to a length three times greater than even the longest cracks observed in as-fabricated devices. It is possible that cracks which formed in this device as a result of processing were long enough to encounter strain in the AlGaN layer induced by electrostatic stress which allowed them to grow. This stress is highest under the gate contact of the device, and several cracks extended to this contact or beyond it.
Given that some of these cracks have bridged nearly the full length of the channel region, it seems plausible that those channel cracks which were present under the gate electrode allowed it to short out the AlGaN locally when it refilled the void formed by each channeling crack. Such a situation would fit with an increase in the gate current of the device, which has been observed previously by other authors studying microcracks with Raman spectroscopy [84], and given that no pitting defects were observed in the channel of the device.
Figure 3-1 A HAADF-STEM image of the cross-section of a 100nm gate length device. The Gate, Source an Drain electrodes are prominently labeled as well as the various material layers present within the device.

Figure 3-2 A metal inclusion formed after the annealing of a Al/Ni/Ti/Au metal stack. These ohmic contacts were formed via an anneal at 850°C for 30s  A) A HAADF-STEM micrograph of the interface reveals the presence of a metal inclusion. B) An EDS of the inclusion suggests that it is formed from TiN.
Figure 3-3  Top-down SEM analysis of the ohmic contact regions of an AlGaN/GaN HEMT. These ohmic contacts were formed via an anneal at 850°C for 30s. A) The wholesale formation of metal inclusions throughout the contacted surface. Cracks can be seen nucleating on the faceted corners of the metal inclusions. B) Cracks which nucleate at the edges of the ohmic contact regions can extend into the channel for much longer distances.

Figure 3-4  FIB/TEM of a nanocrack observed in SEM. A) A lamella for TEM analysis was formed perpendicular to the propagation direction of this crack. B) HAADF-STEM analysis coupled with electron diffraction reveals that the crack extends through the AlGaN layer and is roughly perpendicular to the prism directions.
Figure 3-5  Histograms of crack lengths observed in 20 HEMT devices. A) The crack length distributions associated with cracks found under the ohmic contacts as well as in the channel of the device are shown. B) The tensile stress required to induce additional crack growth for given crack lengths is also shown.
Figure 3-6 Stepped stressing of an AlGaN/GaN HEMT and resulting crack formation. A device was step stressed from $V_{GS} = -10$V to $V_{GS} = -42$V at -1V/min and with $V_{DS} = 15$V resulted in degradation of the device. A) A reduction in the saturation current was observed as a result of this stressing and is represented by the $V_{DS}$-$I_{DS}$ family of curves. B) An increase in the gate leakage current is also observed and is represented by the $V_{GS}$-$I_{GS}$ characteristics of the gate contact.
Figure 3-7 The resulting crack distribution and associated tensile stress for crack growth in the stressed HEMT. After stressing, this device was deprocessed and the distribution of nanocracks observed on its surface was determined. A) This distribution of cracks can be used to estimate the biaxial tensile stress in the AlGaN which used to grow these longer features. B)
CHAPTER 4
PROGRESS IN THE ANALYSIS OF THE RELIABILITY OF THE GATE ELECTRODE

A common factor affecting the overall reliability of AlGaN/GaN HEMTs during DC stressing in the field is the degradation of the gate electrode of these devices. This degradation generally manifests itself as a dramatic increase in the reverse biased leakage current through the gate contact observed as devices are stressed at progressively higher electric fields.

The structural change which is generally accepted as the underlying cause of this increase in current is some reaction which induces shorting of the gate contact to the 2DEG present below the AlGaN epitaxial layer. This shorting can be caused by the formation and refilling of cracks under the gate electrode, metal diffusion down threading dislocations leading to an observed increase in trap centers or even complex electrochemical reactions between the gate contact, the operating ambient, and the semiconducting epitaxial layers of the HEMT device.

This chapter will detail the progress made in understanding the mechanisms of formation associated with this defect in several different HEMT devices. It will begin with an analysis of early work which was performed on Pt-gate devices, where the relatively unreactive metal which forms the gate contact leads to degradation only at very high fields. In these systems, the large inverse piezoelectric strain can lead to the formation of mechanical defects under the gate contact which short out the 2DEG. This section will be followed by a discussion of the degradation of HEMTs which possess a nickel gate electrode. Because of nickel's enhanced reactivity in comparison to platinum, these devices fail at lower fields than their noble metal contacted counterparts, but this failure is much more gradual and progresses by a completely dissimilar mechanism. This chapter will be concluded with a section detailing the efforts made in surface characterization of various HEMT devices with specific attention paid to the observation of defect formation under the gate and its stochastic nature.
4.1 Inverse Piezoelectric Strain and the Reliability of Pt-Gate HEMTs

When an electric field is applied to a highly polar crystal, the bonds of this crystal orient themselves in the direction of the electric field, resulting in the presence of uncancelled charge at the interfaces of the crystal. This reorientation is what ultimately gives rise to the 2DEG in an AlGaN/GaN HEMT [74].

When mechanical strain is applied to a polar crystal, this pushes the alternating atomic planes which give rise to polar bonding closer to one another, increasing the induced electric field within the crystal and the number of charges at the interfaces of the crystal. This electronic response to strain is called the Piezoelectric Effect. A less common phenomenon is the application of strain to a crystal as the result of an applied bias which changes the electric field. This is called the Inverse Piezoelectric Effect. [85].

The strain experienced by a crystal may be expressed with the following pair of coupled equations, where $S$ is the vector representing the strain in all directions applied to the crystal. The symbol $Y$ is the tensor representing the elastic modulus in all directions, while $\sigma$ is the vector representing the stress in all directions (in cm/cm) within the crystal (in Pa). The symbol $d$ is the tensor which represents the inverse piezoelectric coefficients (in pm/V) and $F$ is the vector form of the electric field (in V/cm) [86].

$$\{S\} = [Y]\{\sigma\} + [d]{\{F\}} \quad \text{(4-1)}$$

Because the majority of the field is present between the gate electrode and the 2DEG, the AlGaN layer of the HEMT tends to experience the most stress. For a relaxed system, the first term in the equation can be ignored, as no external stresses are applied to the crystal which would induce strain.

This is not the case for AlGaN/GaN epitaxial structures. Any lattice mismatch associated with epitaxial growth manifests itself in the first term in the above equation as the AlGaN layer
is not thick enough to induce relaxation through defect formation. This tensile stress may be calculated by comparing the "a"-spacing of relaxed Wurtzite Al$_{0.28}$Ga$_{0.72}$N (3.17 Å) to that of pseudomorphic Al$_{0.28}$Ga$_{0.72}$N (3.19 Å) and multiplying the normalized spacing difference between these two materials by the Young's Modulus. Current literature suggests that the strain resulting from this mismatch should be approximately equal to 1.94 GPa [82].

Stress in the metal and passivation layers caused by thermal expansion coefficient mismatches or non-ideal deposition conditions could also result in stress on the AlGaN layer. In order to calculate the stress resulting from the mismatches associated with thermal expansion, a first order model of the gate stack may be employed where a film stack comprised of the 2μm GaN layer, the 14 nm AlGaN epitaxial thin film, and the deposited metal gate (20nm of Ni followed by approximately 500nm of Au) are considered. In this model, the Young's Modulus and thermal expansion coefficients of the Au and Ni (79 and 200 GPa for Young's Modulus, respectively; 1.4x10$^{-5}$ and 1.3x10$^{-5}$ for the thermal expansion coefficient, respectively) are taken from standard texts [87], while the Young's Modulus of GaN and AlGaN are assumed to be roughly equal to 309 GPa [81]. The thermal expansion coefficient for GaN is taken from literature (5.6x10$^{-6}$), while the thermal expansion coefficient of AlGaN is derived from linear interpolation between GaN and AlN (5.6x10$^{-6}$) [88].

The solution for the stress in this thin film is derived by taking the difference between the strain in the AlGaN due to thermal expansion and the "average strain" ($a_0$) of the multistack system, computed as a weighted average, as follows (where $T_1$ is the starting temperature, $T_2$ is the ending temperature, $t_i$ is the thickness of a given film, $E_i$ is the Young's Modulus of that film and $\alpha_i$ is the unitless thermal expansion coefficient of that film) [89]. This "average strain" is shown below.
The resulting stress due to thermal expansion coefficient mismatches derived from this "average stress" is minimal. For an operating temperature of 500ºC, which is large compared to literature results, the computed stress due to mismatch is equal to only 60 MPa of biaxial tensile stress. Given this result, the effects of thermal expansion can be neglected.

Because of the relative magnitudes of the inverse piezoelectric coefficients, AlGaN responds significantly to vertical electric fields between the gate and 2DEG. This response far outpaces stress due to thermal expansion coefficient mismatches, but is not as pronounced as the stress induced by lattice mismatch in the AlGaN. The basal plane is stretched along the prism directions with increasing vertical electric field, putting the AlGaN layer in tension. Particularly high electric fields could, conceivably, cause mechanical damage. In order to calculate the biaxial tensile stress induced under the gate due to the inverse piezoelectric effect to a first order, the following equation is used, where \( d_{31} \) is the biaxial component of the inverse piezoelectric tensor which interacts with the vertical field (assumed to be equal to approximately 1.57pm/V in AlGaN, based off of linear interpolation between GaN and AlN) [90], and \( c_{31} \) is the elastic stiffness constant (assumed to be equal to 103GPa by assuming that the elastic stiffness constant for AlGaN is similar to that of GaN, which seems reasonable given that they share very similar Young’s Moduli) [91].

\[
\sigma = d_{31} c_{31} \left( V_{GS} / t_{AlGaN} \right) \tag{4-3}
\]

The resulting stress from this interaction is 179MPa at 20V, which is 9% of the stress present in the AlGaN film as a result of the formation of a pseudomorphic layer of AlGaN on top of GaN. This is a substantial amount of biaxial stress and it stands to reason that piezoelectric...
stress could induce a fracture event within the AlGaN layer if the conditions were right (perhaps at a sharp corner, like the edge of the gate).

Joh, delAlamo and coworkers have posited that the inverse piezoelectric effect and the strains associated with this effect might lead to the formation of crystalline defects within the AlGaN layer. As these defects form and agglomerate, they induce deep level traps which allow for current conduction between the gate contact and 2DEG through the AlGaN. As these defects continue to agglomerate and as the stress increases, plastic deformation and crack formation may occur, resulting in more current conduction. They further posited that this defect formation may be the cause of substantial degradation observed in stressed HEMT devices which are stressed in conditions where a substantial lateral field is not present to aid in hot-electron based degradation [92].

If the inverse piezoelectric effect is the cause of defect formation during off state stressing, the degradation should depend upon the vertical field present in the AlGaN semiconductor underlying the gate electrode. There should also exist some critical voltage at which a substantial change in the gate leakage current occurs, as mechanical defects form. This critical voltage should also change with any mechanical stress applied to the HEMT. Increasing the tensile stress present within the HEMT should reduce the critical voltage while increasing compressive stress should increase the critical voltage required for defect formation. Finally, because the inverse piezoelectric effect should be highly dependant upon the field present within the device, defects should form preferentially on the drain side of the gate.

Joh and coworkers stressed a group of devices at $V_{DS}=0\text{V}$ with $V_{GS}$ stepped from -10V to -50V at a rate of 1V/s on an AlGaN/GaN HEMT with a Pt gate at room temperature and in an N$_2$
ambient. The results of this off-state stressing was a substantial increase in reverse bias leakage current accompanied by a moderate increase in the forward biased current as well.

This measurement was repeated at variable temperatures both before and after stressing in order to determine an activation energy for the increase in the forward and reverse currents characteristic to these rectifying contacts. The activation energy was determined by fitting the changing diode current density, \( J(T) \), to the following equation, where \( J_0 \) is a pre-exponential constant, \( k \) is Boltzmann's constant, \( T \) is the temperature (in Kelvins), and \(-E_A\) is an activation energy (in eV/K).

\[
J(T) = J_0 \left( e^{-E_A/kT} \right)
\]  

(4-4)

The activation energy associated with forward biasing did not change from its starting value of 0.26eV as a result of stressing, indicating that the degradation observed does not result from an alteration of the built in voltage of the device. However, the value of the activation energy of the reverse current changed from a value of 0.45eV to a value of 0.003eV, indicating that some event occurred during stressing which fundamentally altered the mechanism of leakage through the gate contact.

The stressed HEMT device also demonstrated substantial reductions in \( I_{DS} \) after degradation as well as a reduction in transconductance. Because stressing at \( V_{DS}=0 \) results in minimal current through the channel in comparison to "on" state stressing, the collapse in \( I_{DS} \) and reduction of transconductance cannot be ascribed to hot electrons. This degradation was attributed in some part to an increase in the drain resistance, which fits well with a field effect being the cause of degradation [93].

The leakage current remained low but increased exponentially with \( V_{GS} \) below a critical voltage (\( V_{CRIT} \)). This was followed by a range over which this exponential increase in gate
current occurred at a much higher rate and resulted in an increase in current of over two orders of magnitude. After this region of rapidly increasing gate leakage current, the exponential rate of increase returned to its original value. This experiment was repeated in the "on" state, with a variable I_{DS}. The same critical voltage is observed as V_{GS} was stepped from low to high values, but the critical voltage did not appear to vary in any rational manner as I_{DS} changed.

When stressing occurs in "off" mode (V_{DS} no longer equal to zero) V_{CRIT} decreases as V_{GS} and the associated maximum vertical electric field density under the gate increases. This was demonstrated by Joh and coworkers in an experiment where they applied varying potentials between the gate and source electrodes and proceeded to step the potential at 1V/min between the gate and drain from values corresponding to V_{DS}=0 up to -50V [95].

Another experiment with V_{DS}=0 was performed on HEMTs with gatelengths that varied from 1.50μm to 0.25μm. V_{GS} was stepped at a rate of 1V/s from 0V up to a value of 50V. As was predicted by Joh and coworkers, the value of V_{CRIT} decreases with decreasing gate length and increasing maximum electric field. This decrease appears to be particularly substantial at aggressively scaled gatelengths [96].

As shown schematically in Figure 4-3, In order to validate the cause of the degradation observed in the studies described previously, Joh and coworkers performed pulsed stressing on the gate electrode of a Pt-gated HEMT with V_{DS}=0. This pulsed stressing consisted of three cycles. In each cycle, there is a two hour stressing period, where the gate of the device is subjected to a potential of -40V and degradation of the HEMT should occur, followed by a one hour recovery period where the gate of the device is set to a value of 0V. At each minute of the measurement, the gate current in "off" mode, as well as the gate and drain currents in "off" mode were measured. Because the stress voltage was of a higher magnitude than V_{CRIT}, off-mode I_G
rapidly increased over the first stressing period and stabilized at its maximum value for each stressing period. On-mode $I_D$, however, degraded exponentially over time during each stressing period. When the device cycled into recovery, current collapse was reduced as $I_D$ increased at a rate corresponding to a sum of inverse exponentials. It returned to its pre-recovery value whenever a new stressing cycle began and the maximum current achieved at the end of the recovery period decreased during each cycle. This behavior is consistent with trap formation in the AlGaN layer which decreases the conductivity of the 2DEG and causes a recoverable current collapse as a result [97].

As shown in Figure 4-4, Joh and coworkers have documented the formation of pitting along the drain edge of the gate electrode during on-mode stressing ($V_{DS}=40V$, $I_G=250mA/mm$): a phenomenon consistent with a degradation process caused by the inverse piezoelectric effect. These pitting defects were evaluated using cross-sectional transmission electron microscopy (XTEM) [98] as well as scanning electron microscopy (SEM) and AFM of a surface deprocessed with a combination of wet chemical treatments, which will be described in greater detail later in this work.

With this method, Joh and coworkers were able to observe the formation of metal-filled pits on the drain edge of the gate electrodes of Pt-gated AlGaN/GaN HEMTs stressed in "off" mode ($V_{GS}=-7V$) with $V_{DG}$ stepped from 8V to 50V at a rate of 1V/s. These pits grow with increasing gate to drain voltage when the HEMT is biased in "off" mode, which is consistent with some event driven by the electric field. The leakage through the gate appears to follow the density of pitting defects [99].

The results above indicate that the degradation of devices in "off" mode, "on" mode and when $V_{DS}=0$ is driven by the vertical field present in the AlGaN layer and not by current density
through the 2DEG, as is the case for hot electron effects induced in the "semi-on" mode. The results do not rule out, however, that the gate current density, rather than the density of current in the 2DEG might influence the reaction, so a positive correlation with the inverse piezoelectric effect cannot necessarily be assumed for Pt gate HEMTs.

A recent detrapping study performed by Kuball and coworkers on a Pt-gate HEMT possessing a Al_{0.26}Ga_{0.76}N/GaN epitaxial stack with no capping n-GaN layer does suggest that the defect which forms is related to metal diffusion. This HEMT was stressed in "off" mode (V_{DS}=30V, V_{GS}=-5V) from room temperature to 150°C and for times ranging from 0h to 40h. Detrapping analysis consisted of a voltage-driven measurement where the sample was put in a "filled" state (V_{GS}=-10V, V_{DS}=0V) for 1s (in order to induce trap filling) and an "transient" state (V_{GS}=1V, V_{DS}=.5V) for 1000s (to observe trap emptying) [100,101]. This methodology is demonstrated, schematically, in Figure 4-5.

Kuball and coworkers observed the presence of a band of traps which increased dramatically over the course of stressing and progressed from a peak trap energy of 0.45eV to 0.65eV. The magnitude of this trapping signature, C (in cm\(^3\)), (which should, itself, be linearly related to the trap density) indicates that the trap concentration varies with respect to the following equation, representing one-dimensional diffusion from a pseudo-infinite source [102].

\[ C(x, t) = C_0(x, t)\left(1 - \text{erf}\left[z/(2D t_{\text{AlGaN}})^{1/2}\right]\right) \quad (4-5) \]

In this equation, z is the characteristic tunneling length from the 2DEG into the AlGaN layer (3nm), D is the diffusivity (in cm\(^2\)/s), of diffusing, defect-forming species (likely Pt) down into the AlGaN layer, t_{\text{AlGaN}} is the time in seconds and S is the surface concentration (in cm\(^3\)) of defect forming species. Variation of the observed diffusivity with temperature was fitted with an Arrhenius Function, as seen in the equation below, where D_0 is the diffusivity at an infinite
temperature (in cm$^2$/s) and $E_A$ is the activation energy (in eV) for diffusion and $k$ is Boltzmann's constant.

$$D = D_0 \left( e^{-\frac{E_A}{kT}} \right)$$

(4-6)

The activation energy for diffusion was found to be roughly equal to 0.23eV. This corresponds well with previous studies of oxygen diffusion down dislocation lines during annealing by Pearton and coworkers [103]. Kuball and coworkers suggest that this indicates that pitting-related defects form via metal diffusion down pre-existing threading dislocation lines under the gate. This diffusion is enhanced by piezoelectric strain induced by the vertical field. It bears noting that this process is similar to degradation processes observed in GaAs HEMTs formed in the early years of research into the material, when dislocation densities were large [104,105].

A few general trends detailed above for Pt-gate HEMTs also hold true for other gate technologies and bear special emphasis. Firstly, the sharp increases in gate current observed during stepped stressing of $V_{GS}$ are commonly observed in other gate-stack technologies. The voltage at which this transition from low leakage current to high leakage current occurs is variable from one technology to the other. Furthermore, the manner in which gate current increases, or whether it increases at all, after the onset of the critical voltage varies from one architecture to the next [106].

The increases in observed drain resistance as a result of gate shorting is also common among various gate-stack technologies. This effect is induced by the formation of an alternative path for current to travel out of the device, that being through the gate rather than through the drain. This path for leakage current reduces the observed current through the drain of the device, for a given voltage, resulting in a lower observed resistance via Ohm's Law.
Lastly, it bears noting that Pt-gated HEMTs are not the only architecture where the gate leakage current has been correlated with the formation of some form of defect bridging the gate and the underlying semiconductor material. In Pt-gated devices, this sort of defect is induced by mechanical failure, specifically cracking, and refilling with gate or passivation material. In other devices, however, electrochemistry or diffusion may also be involved. It is generally agreed upon, however, that the electrostatic field present between the gate electrode and the 2DEG drives any reactions responsible for the degradation of the gate electrode.

4.2 Reliability of Ni-Gate HEMTs

Initial studies of the failure of the Ni-gate electrode of an AlGaN/GaN HEMT was performed by Chang and coworkers on devices similar to those which will be used in these studies. A device with a 140nm gatelength AlGaN/GaN HEMT with $V_{GS}$ step stressed from -10V to -42V at -1V/s and $V_{DS}$ held at 5V. The failed device exhibited the same sharp increase in gate leakage current in off-mode as was the case for Pt-gated devices studied previously by delAlamo and co-workers. This non-recoverable spike in gate leakage current is also evident in the plot of $I_{GS}$ versus $V_{GS}$ for the HEMT device, which demonstrates that the reverse biased leakage current in the gate contact of the device is dramatically increased after stressing [107]. This is also similar to results observed by delAlamo and coworkers.

Curiously, the observed gate leakage did not stabilize after this initial increase after $V_{CRIT}$, as was the case in Pt-gated HEMTs. Instead, the observed off-state leakage current continued to increase. It also bears noting that as a general trend in these devices, depending upon the stressing conditions, the relative increase in gate current observed after exceeding $V_{CRIT}$ is highly variable. Generally, the relative increase in gate current at $V_{CRIT}$ increases with increasing voltage between the source and drain electrodes of a Ni-gated HEMT [108]. Both of these observations suggest that the mechanism of degradation for a HEMT device possessing a
nickel gate may be fundamentally different from the mechanism of degradation of Pt-gated HEMTs. An example of a stressing experiment similar to the one described in Chang's work is demonstrated in Figure 4-6.

Chang and coworkers also observed a decrease in the intensity of the peak in the photoluminescent spectrum correlating to the band-to-band radiative recombination event for GaN after stressing. This decrease in the PL intensity observed after stressing suggests the formation of non-radiative recombination centers as a result of off-state stressing, which would quench the band-to-band, radiative recombination peak. The relative decrease in peak intensity was maximized near the gate and drain electrodes, which correlates well with the expected position of the degradation present within the device.

It is often the case that a native oxide is formed on the surface of the AlGaN layer of a HEMT after all epitaxial layers have been deposited and vacuum is broken to move the device from epitaxial processing equipment to lithography and metallization. In the devices studied as part of Chang's work as well as in this work, this interfacial native oxide layer is approximately 1.5nm thick. XTEM of the gate electrode of Chang’s Ni-gated device did not demonstrate the existence of a crack at the drain side of the gate electrode, as was observed previously in Pt-gate devices. Instead, intermixing and dissolution of the interfacial layer was observed.

Douglas and coworkers observed decreases in $V_{CRIT}$ in HEMTs which used a Ni gate rather than the Pt gates studied by Joh and coworkers. The gate length of the Ni electrode was scaled from 100nm to 170nm and the devices were tested at $V_{DS}=0$ with $V_{GS}$ stepped at a rate of 1V/min from -5V to -45V. While the critical voltage does change dramatically with HEMT gate length, as was demonstrated previously by delAlamo and coworkers, the simulated critical vertical electric field within the device did not appear to increase, remaining fixed at
approximately 2.5 MV/cm², further suggesting that degradation is induced by the vertical electric field within the HEMT. In additional studies, where stepped stressing of \( V_{GS} \) from -10V to -42V was performed on devices at a variable \( V_{DS} \) ranging from 0V to 15V, Douglas and coworkers observed that the critical voltage at which a Ni-gated HEMT degrades is related to the voltage difference between the drain and source electrodes as well as the voltage difference between the gate and source electrodes, with devices stressed at a higher \( V_{DS} \) having a lower critical \( V_{GS} \). However, if the critical voltage for failure was analyzed in terms of the voltage between the gate and the drain rather than the gate and the source, \( V_{CRIT} \) remained fixed at a similar number as was observed in the case of stepped stressing at \( V_{DS}=0V \) [109]. The field which resulted between the 2DEG and the gate of the HEMT was comparable to the field calculated in the case of stepped stressing with \( V_{DS}=0V \), suggesting that the field present between these two high-conductivity layers is the driving force for degradation in Ni-gated HEMTs.

Using Equation 13, in a similar fashion to how it was described previously, a field resulting from a voltage of -22V (the approximate critical voltage) results in a value of 197MPa. This value can be related to the fracture toughness of a 2D channeling crack and, through it, the surface energy of AlGaN via another equation, similar to Griffith's Equation [110]. It shares all the same variables (even \( Z \) is the same – 1.976), except one, which is \( h \): the depth of the channeling crack.

\[
\sigma = \left[ \frac{(2E\gamma)}{(Zh[1 - v^2])} \right]^{-1/2} \tag{4-7}
\]

This is not enough stress to induce 14 nm of crack growth, which is, at minimum, 5GPa. It stands to reason that fracture is not a direct driving mechanism for defect formation in these devices. It also bears noting that Douglas and coworkers performed additional studies of Ni-gated HEMT failure at voltages lower than \( V_{CRIT} \) in an unpublished study. Surprisingly, they
discovered that the spike in leakage current which occurs at the observed critical voltage during stepped stressing can also occur at voltages much less than $V_{\text{CRIT}}$. This time-dependant critical voltage suggests, firstly, that degradation in stepped-stressing is not dependant only on the field, but is also dependant upon time. This implies that the two factors may be convolved in stepped stressing experiments, making analysis difficult. The range of voltages at which this spike in leakage current is observed is so broad that shorting of the AlGaN due purely to cracking seems unlikely, as well. Overall, the preponderance of data suggests that some mechanism other than strain-induced material failure at the gate electrode is the primary driving force for degradation at $V_{\text{CRIT}}$ and beyond.

Evidence exists that degradation kinetics are different in HEMTs which utilize nickel gate materials. Lo and coworkers compared the degradation of Ni-gated and Pt-gated Al$_{0.25}$Ga$_{0.75}$N/GaN 1μm gate length HEMTs. These HEMTs were stressed at $V_{\text{DS}}=5$V with $V_{\text{GS}}$ varying from -10V to -100V. HEMTs possessing a Ni gate appear to degrade at a much faster rate, at a lower critical voltage and to higher absolute leakage currents than the Pt gated HEMTs. Also, current collapse is more pronounced in these devices than in Pt-gate HEMTs. Lo and coworkers also evaluated the X-Ray Photoemission Spectroscopy of as-formed Ni and Pt gate contacts on GaN as well as similar contacts which had been annealed for 30 min at 300°C. XPS indicated no change in bonding configuration for the Pt gate electrodes. However, it indicated that, during annealing, the Ni appears to diffuse into the GaN semiconductor. The oxygen present in the interfacial GaO layer reacts with the Ni preferentially to form nickel-oxygen bonds [111].

A similar effect was noted by Burnham and coworkers, who stressed Ni-gate AlGaN/GaN HEMTs in "on" mode with $V_{\text{DS}}=15$V and $V_{\text{GS}}$ set in order to achieve the $I_{\text{DS}}$
corresponding to a channel temperature which varied from 136°C to 320°C and which was stepped at a rate of 23°C/day. Both samples were stressed until an 11% collapse in $I_{DS}$ was observed. HEMTs which were stressed in an air ambient failed much more rapidly and at a lower temperature (205°C and 100h) than HEMTs which were stressed in an N$_2$ ambient, which failed at a temperature of 320°C after over 350h of stressing. They observed with XTEM that a defect had manifested itself over the full length of the gate contact in both devices rather than only at the drain electrode. They observed nickel interdiffusion into the semiconductor along the full gate lengths of devices stressed in both air and N$_2$, not just at the drain side of the gate edge as had been noted earlier in the pitting studies by Joh and coworkers. A substantial oxygen content was observed in devices stressed in an air ambient, suggesting oxygen diffusion and reaction with the interdiffusing nickel [112].

Holzworth and coworkers performed XTEM on a Ni-gated devices failed in an oxygen ambient at room temperature. The devices were stressed at $V_{DS}$ of 5V while $V_{GS}$ was stepped from -5V and -10V to -42V at -1V/min increments. They observed that samples degraded under these conditions possessed the characteristic decreases in $I_{DMAX}$ and increases in $I_{GS}$ which corresponded to defect formation in previous studies of defects Ni and Pt gated electrodes. They further observed that devices which did not degrade rapidly and which did not experience significant drops in $I_{DMAX}$ also did not show evidence of defect formation in XTEM.

Devices which did degrade at an accelerated rate and which manifested dramatic decreases in observed $I_{DMAX}$, however, did yield evidence of defect formation during analysis with XTEM. These defects were of a profoundly different nature than the defects previously observed in XTEM of Pt-gated devices by delAlamo and coworkers [113]. As shown in Figure 4-7, the defect which formed as a result of off-state stressing is not confined to the drain-side of
the gate contact of the device, as was previously observed. The gate contact of the degraded
device was also pushed out of contact with the AlGaN epitaxial layer, as well as with the nitride
passivation layer which had covered its surface, by well over 20nm. This defect also spanned the
full thickness of the AlGaN, resulting in direct shorting of the gate electrode of the device to the
2DEG.

The defect, itself, was determined to be amorphous via analysis with in situ electron
diffraction. EELS mapping of the defect observed in cross-section demonstrated that the defect,
itself, was comprised of a mixture of nickel and oxygen. The shape of the defect itself was
commensurate with the electric field lines present in the HEMT device and modeled with the
Florida Object Oriented Device Simulator (FLOODS).

A variety of conclusions about the mechanism of formation for pitting defects in Ni-
gated HEMTs can be drawn from the observations made in this study. Firstly, it is apparent that,
unlike degradation in Pt-gated devices, the formation of the defect under the gate is not driven
purely by cracking due to piezoelectric strain and that some electrochemistry must be involved
which gives rise to the formation of a nickel-oxide inclusion. This inclusion is, volumetrically,
larger than the nickel liner layer which it formed from and it pushes the gate electrode out of
contact with the AlGaN layer. This defect still appears dependant upon electric field in some
capacity, as its morphology matches closely with the magnitude of vertical field present between
the gate electrode and the 2DEG of the device. This correlates well with other studies of Ni-
gated AlGaN/GaN HEMTs, which have demonstrated that changes in device architecture which
can reduce the residual electric field around the gate, such as the introduction of a field plate, can
greatly enhance the performance of devices during stressing, resulting in a much higher $V_{CRIT}$
than is achievable without the aid of a field plate [114].
4.3 Efforts in Surface Characterization of AlGaN/GaN HEMTs

While efforts to characterize device degradation with cross-sectional TEM have met with success in both Pt-gated and Ni-gated HEMT architectures, efforts to quantify defect formation with these techniques have not been as effective because TEM methods are not as sensitive to stochastic effects, as has been described previously.

DelAlamo and coworkers developed experimental methods to better analyze the stochastic effects which play a part in the degradation process by utilizing a combination of novel etch chemistries as well as top-down SEM as well as AFM measurements. Their method begins with deprocessing of the device structure in order to expose the epitaxial AlGaN layer to inspection with SEM or AFM. This deprocessing begins with an exposure in HF, meant to eliminate the passivation nitride which sits atop of the metal layers present within a device as well as the AlGaN which comprises the channel.

This exposure in HF is briefly followed by an exposure in Aqua Regia, meant to eliminate the Au alloyed and Pt alloyed ohmic and gate metal contacts, respectively. Aqua regia was chosen over a FeCN/KI based chemistry, described earlier, likely because of its effectiveness in etching platinum metal layers, which FeCN/KI is incapable of accomplishing. This aggressive metal etchant does have its draw-backs, however. Aqua regia does not exhibit the high selectivity of FeCN/KI between metal layers and the underlying AlGaN. Rather, it tends to attack the AlGaN, resulting in the formation of etch pits and other surface topographies which can result in false-positives when pits are being counted, measured and analyzed.

Despite this drawback, this Aqua Regia based deprocessing solution has been used to great effect in studies of Pt-gated as well as Ni-gated HEMTs [115]. Makaram and coworkers used this methodology in conjunction with SEM to identify the formation of crack-based pitting in Pt-gated devices and to find a direct relationship between the pit densities with the leakage
current observed in these devices. They also made use of AFM to demonstrate that the size of crack-induced pits increases with increasing applied reverse bias to the gate electrode of these Pt-gated devices, indicating that degradation of the device continued despite the fact that large increases in the gate leakage current is generally not observed in Pt-gated devices.

The deprocessing techniques used by Makaram and coworkers have found application in HEMTs with nickel gates. Gao and coworkers used the same deprocessing method as was described previously to analyze pit formation in Ni-gated HEMT devices with a 1µm gate length in an air ambient [116]. These devices were biased with $V_{DS}$ equal to 5V and $V_{GS}$ stepped from -5V to -40V. This biasing scheme resulted in an asymmetric electric field. In these devices, the field was highest between the gate electrode and the 2DEG, specifically at the drain side of the device, where the difference in potential between the gate and 2DEG was maximized. Not surprisingly, this device manifested the same gate leakage current dependance on stepped gate-to-source voltage as had been observed in devices previously, with a sharp increase in current at $V_{CRIT}$.

Deprocessing by Gao and coworkers revealed that the increase in gate leakage current during stepped stressing was induced by pitting of the AlGaN layer under the gate electrode. Furthermore, this pitting was closely confined to the drain-side of the gate electrode, where the field was maximized. They further demonstrated that the physical location of the pitting under the gate could be switched from one edge of the gate to the other if the biasing scheme was changed such that the source of the device was biased at the higher voltage than the drain (effectively making the source the drain and the drain the source). This further illustrated that electric field has a major impact on the reactivity of the gate metal with the underlying AlGaN/GaN as well as the ambient.
The chemical deprocessing methodology described above has also been used in conjunction with non-destructive techniques such as Electroluminescence spectroscopy, to reveal the impact of stepped stressing and associated defect formation under off-state conditions on the current transport through a HEMT. Excitons in AlGaN GaN, like many other compound semiconductors, undergo direct, radiative, band-to-band recombination, where the momentum of the constituent electron and hole of the exciton is conserved and all energy is transferred to a photon. This photon is free to exit the material system and be collected by a sensing instrument like, in the case of experiments done by Bajo and coworkers, an astronomy-grade CCD camera in line with a 50X magnification microscope objective [117].

Bajo and coworkers stressed a single Ni-gated device with 600nm gate length in "off" mode, with $V_{DS}$ equal to 30V and with $V_{GS}$ equal to -15V for a 760s time period and viewed it through the CCD-equipped microscope detailed above. They observed the formation of a group of bright spots of EL intensity between the gate and drain of the device under stress. These bright spots matched closely with the location of pits present under the drain side of the gate of this device which were very shallow (being 2-4nm deep) and very broad (being 100-200nm wide).

In order to understand the correlation between the bright spots which Bajo and coworkers observed with EL and the pitting which they observed with AFM, it is important to consider what leads to variation in EL intensity. The EL intensity is directly proportional to the number of direct recombination events which occur in a given region of a material. Not surprisingly, the number of direct recombination events which occurs in a given time period and in a given material volume is related to the product of the concentrations of electrons and holes in the
material as well as a general "attempt frequency". More EL intensity arises from areas where more electrons are present in the material and, therefore, more recombination events occur.

In the device observed by Bajo and coworkers, the path of least resistance for leakage current passing through the device was from the drain of the device, through the drain-side of the channel and, then, through the areas of the gate where pitting had occurred. So, electron current in the 2DEG crowded around these pitted regions. The random recombination events where these crowded electrons recombined with holes present in the GaN resulted in the formation of luminescent "spots" in the EL image. Bajo's observation of these "spots" in EL suggests that leakage current through the gate electrode is related to the presence of pitting defects and, as is the case with Pt-gated devices, that the leakage current increases with increasing pit density.

All of the research above paints a very clear picture of the degradation which appears to occur in a Ni-gate HEMT after long periods of stressing or when steppe stressing occurs up to potentials well in excess of the critical voltage. Very little materials characterization has occurred at the critical voltage, itself, however. This might warrant further study.
Figure 4-1 Piezoelectric strain (white arrows) resulting from a vertically applied electric field (black line) in an AlGaN/GaN HEMT.

Figure 4-2 The dependance of gate leakage current with increasing electrostatic stress. The electrostatic stress was applied between the gate and the source electrodes at a constant rate ($V_{GS}=-10V$ to $V_{GS}=-42V$ at -1V/min). The regions of potential lower than $V_{CRIT}$, equal to $V_{CRIT}$ and greater than $V_{CRIT}$ are highlighted.
Figure 4-3  Pulsed stress experiments on an AlGaN/GaN HEMT. 3h stressing periods at $V_{DS}=0V$ and $V_{GS}=-40V$ were followed by a 1h detrapping periods to determine the effect of traps at the gate-AlGaN interface on the current collapse and gate leakage current [92].

Figure 4-4  A schematic image of a crack-like defect formed at the drain side of the gate. This sort of defect could form of an AlGaN/GaN HEMT stressed at $V_{DS}=40V$ with $V_{GS}$ set to yield $I_{DS} = 250mA/mm$ [98]. The crack-like defect formed on the drain side of the gate and refilled with passivation dielectric, silicon dioxide or gate metal, depending upon the defect imaged.
Figure 4-5 The transient electrical measurement performed by Kubal and coworkers. This measurement was performed on AlGaN/GaN HEMTs prior to and during stressing to measure traps present in the device. This measurement began with a UV light pulse, meant to empty all traps present within the device, followed by a 1s trap filling pulse on the gate electrode while the drain and source are shorted to ground. An emission transient is observed when the device is switched into "on" mode.
Figure 4-6 An exemplary stressing experiment performed on a Ni-gated AlGaN/GaN HEMT. In this case, $V_{DS} = 5\text{V}$ and $V_{GS}$ was stepped from $-1\text{V}$ at $-1\text{V/min}$. The increase in gate current at the critical voltage, as well as the continuing increase in gate current beyond this voltage, is pictured.

Figure 4-7 BF-TEM image of a defect under the gate of an AlGaN/GaN HEMT. The electrochemical reaction of nickel and the subsequent consumption and pitting of the AlGaN surface results in an enhanced leakage current through the gate.
As was discussed in the previous chapter, stepped stressing has become one of the most popular techniques utilized in understanding the nature of degradation in AlGaN/GaN HEMTs. It is utilized by many different organizations and has been used to examine failure mechanisms on many different device architectures. While the technique has been used to great effect in understanding the operation and degradation of Pt-gated HEMT structures in the field, stepped stressing is only just beginning to be utilized extensively in the study of failure in Ni-gated devices.

Research performed previously by various authors has demonstrated that the failure of these Ni-gated devices is unlike what has been observed previously in HEMTs formed with a Pt gate, where cracking induced by the inverse piezoelectric effect seems to be the driving force for defect formation as high fields on the drain side of the gate induce the formation of grooves and cracks which can refill with passivation material or gate metal itself, shorting out the gate to the 2DEG. In Ni-gated devices, this same shorting event occurs (and appears to be heralded by a spike in gate leakage current at $V_{\text{CRIT}}$), but it does not appear to be induced by cracking between the 2DEG and gate contact of the device. Rather, a new defect forms, comprised of some mixture of nickel and oxygen as well as trace concentrations of aluminum and gallium.

The formation of this defect appears to be mediated by electric field, given that the critical voltage required to induce a large increase in gate leakage current increases for larger gate lengths, maintaining the critical field required for defect formation (as is the case for Pt-gated devices). However, defect formation also appears to be mediated by the ambient in which the stressing occurs, with ambients capable of supplying a steady flow of oxygen to the sample inducing failure of the device much more rapidly than inert ambient. Studies performed by
various authors also suggest that the interface present between the metal gate and the epitaxial AlGaN layer of the device plays a major part in defect formation and failure of the device. Mixing at this interface has been posited as a possible cause of increasing leakage current at the critical voltage and the thickness of the interfacial native oxide layer separating the gate metal from the AlGaN has been shown to impact the formation of pitting defects by diffusion. In order to fully understand the various mechanisms which may play a part in defect formation in these Ni-gated devices, the behavior of example devices during various stages of stepped stressing must be studied.

The performance of an AlGaN/GaN HEMT device during stepped stressing can be effectively separated into three separate regimes. The first of these regimes is operation at voltages below $V_{\text{CRIT}}$, where no permanent increase in leakage current is observed. Of the three regimes, defect formation at voltages well above $V_{\text{CRIT}}$ is, probably, the most well documented and well understood. It is degradation within this regime which led to the formation of the nickel-oxide based defects previously observed during stressing of various Ni-gated HEMTs. No direct studies of Ni-gated HEMT devices have been performed where these devices were stressed only up to $V_{\text{CRIT}}$. This regime is, probably, the least well-understood for the case of defect formation in AlGaN/GaN HEMTs.

In order to better understand the degradation and overall change in behavior of AlGaN/GaN HEMTs in these three regimes, three separate devices were stressed in off-mode, with $V_{\text{DS}}=5\text{V}$ and with $V_{\text{GS}}$ stepped at a rate of -1V/min from a starting voltage of $V_{\text{GS}}=-5\text{V}$ to a variable ending voltage with a maximum of $V_{\text{GS}}=-42\text{V}$. The first of these devices was stressed to a maximum potential between the gate and source of $V_{\text{GS}}=-16\text{V}$. This ensured that this device did not exceed the critical voltage, maintaining the device under test in the first low voltage
regime of stepped stressing. The second device was stressed up to a voltage of $V_{\text{GS}}=-16\text{V}$, achieving $V_{\text{CRIT}}$ and ensuring that the second device had entered the second regime of stepped stressing. A final device was stressed up to a voltage of $V_{\text{GS}}=-42\text{V}$, well above the critical voltage and well into the third regime of operation for stepped stressing. A measurement of the gate leakage current passing through the gate electrode of the device as well as the currents passing through the source and drain was made after every stepped increase in voltage. This measurement occurred over a 1ms timespan, so the duty cycle of stressing may be assumed to be roughly equal to 100%.

As mentioned previously, all stressing and measurement of device characteristics was accomplished via the use of an HP4146C device tester, capable of independant measurements of current and voltage on all three device contacts, and a Techtronix 370A curve tracer. The devices used in this study were formed using the processing described in the first chapter of this dissertation on a SiC substrate. All devices stressed as part of this study possessed a gate-length of 100nm. Analysis as part of this initial investigation involved the deprocessing of these three devices after stressing, using the method detailed in the middle section of the experimental chapter of this work. After deprocessing, these stressed devices were analyzed with plan-view SEM (performed on an FEI Strata DB235 Focused Ion Beam Miller in electron control mode using the TLD in UHR imaging mode), which was utilized to quantify the aerial density of defects observed after stepped stressing.

What proceeds from this point onward is a discussion of the behavior of AlGaN/GaN HEMTs under stepped stressing conditions in all three of these regimes, with the experiment described above acting as a means of reference and comparison from which a group of more detailed experiments studying behavior in each of these regimes is derived. The sections of this
chapter will follow directly from the three separate regimes of stepped stressing discussed
previously, starting at low potentials between the gate contact and the 2DEG and proceeding to
ever-increasing potential.

5.1 Device Quality Below the Critical Voltage

As demonstrated in Figure 5-1, if the potential between the gate and 2DEG of the device
is not increased above the critical voltage, substantial increases in gate leakage current do not
occur. This suggests that defect formation does not play a critical role in the characteristics of
the device under stress at these low voltages, and the results from plan-view SEM analysis of
these devices appears to corroborate this supposition. No defects were observed after
deprocessing and SEM analysis of this device. As is evident upon inspection of the
representative image in Figure 5-1, the deprocessed channel is devoid of any features at all
except for some dark spotting indicative of residual organic contamination on the surface of the
sample.

It should be noted that this result is not an indication that defect formation is impossible
at voltages below the critical voltage defined by a -1V/min stepped stressing experiment. As has
been previously observed by Douglas and coworkers, enhanced leakage currents (and,
presumably, the formation of defects under the gate electrode of the device) can be achieved at
voltages below the critical voltage defined by the experiment above, indicating that the rate of
reaction which leads to defect formation is non-zero below $V_{\text{CRIT}}$, as it is defined in this
experiment.

The implication of this information in conjunction with the first result of the experiment
described below is that, while the reaction which appears to induce defect formation is still active
below the critical voltage as it is defined by this experiment, it is definitely mediated by field.
Furthermore, the results of this part of the experiment also indicate that leakage current may be a
good indicator of defect formation under the gate electrode of the device. No defects exist in the absence of enhanced leakage current from the gate electrode of the device.

5.2 Defect Formation at the Critical Voltage

Figure 5-2A demonstrates that the critical voltage, that point where the off-state leakage current passing through the gate electrode of a HEMT device increases dramatically as potential is applied between the gate and the source, is the initial point where defect formation begins. Observation of the leakage current graphed in the first part of the figure demonstrates that as the voltage approaches $V_{\text{CRIT}}$ for the device, the gate leakage current begins to increase. In this case, the leakage current passing through the device increased by fully two orders of magnitude. The image of the deprocessed device after stressing, shown in Figure 5-2B, indicates that defect formation begins as this leakage current begins to increase.

Interestingly, the defect formed in the device in this stressing regime bears little to no resemblance to defects observed previously by delAlamo, Palacios, Kubal, and others. While previously observed pitting defects appeared at random sections along the gate stochastically, this defect appears to have formed along the majority of the entire gate length during stressing. In contrast to typical pitting defect formation, which might have an aerial density of only a few percent in comparison to the total gate area, this defect was present along fully 66% of the full gate width [118].

The morphology of this defect also appears strikingly different from the morphology of previously observed pitting defects. This defect manifests itself as a dark band of contrast in the secondary electron image, extending along the entire gate width. This dark band of contrast is two-toned for the 100nm gate width device. The darker toned portion of the band has a width which matches well with the width of that section of the gate which directly contacts the surface of the AlGaN epitaxial layer (100nm). The extent of the lighter banding to either side of this
dark region matches well with the total width of the top structure of the gate electrode, which does not sit in direct contact with the AlGaN epitaxial region (500nm).

The dark contrast of the band relative to the rest of the channel region of the AlGaN/GaN HEMT may be explained by considering how the secondary electron yield is influenced by different materials. Secondary electrons only have tens of eV of energy because they are the result of inelastic collisions between the bombarding electrons of the SEM and the electron clouds of the atoms which comprise the materials which the bombarding electrons pass through. The yield of these secondary electrons is exponentially dependant upon their depth in the film, because larger depths imply more material which must be travelled through in order to reach the top surface and have a chance of ejection into the vacuum and interaction with the secondary electron detector. Secondary electron yield is also dependant upon the workfunction of the interface between the sample material and the vacuum. Materials with larger workfunctions (such as insulators) will have a dramatically lower secondary electron yield than materials with small workfunctions (such as metals). Thus, it seems plausible that the contrast observed in SEM is due to a thin insulating layer present on top of the AlGaN surface which absorbs electrons on their way to the interface between the sample and the vacuum and which also has a large workfunction which prevents the ejection of electrons out of the sample [119].

It bears noting that this defect could not be removed with an additional round of the HF/TFAC based deprocessing solution. After one hour of exposure in HF, this feature was finally eliminated. The appearance of this banded structure serves as further indication that the mechanism of defect formation during stepped stressing is not the same for Ni-gated devices as for Pt-gated devices studied previously by various authors. Indeed, defect formation in the HEMTs which were analyzed as part of this study appears to be comprised of two separate
processes rather than a single process, as is the case in Pt-gated devices: those being pit formation at high voltages and (at low voltages approaching $V_{\text{CRIT}}$) banding. The nature of this new banding defect is largely unknown, however. In order to understand the mechanism of its formation and the manner in which it degrades the performance of the device, further studies of the composition and formation of this defect are required.

Firstly, it is important to understand the electrical conditions, besides the sudden increase in gate leakage current at $V_{\text{CRIT}}$, which are associated with this banding defect. As shown in Figure 5-3A, the potential applied across the 2DEG between the source and drain electrodes also appears to impact the formation of the banding defect under the gate contact of the device. The devices under study in this case were processed in a similar fashion to the processing scheme described previously, in Chapter 1, with the exception that the architecture was designed to accommodate a circular gate with a field plate attached, rather than a linear gate with no field plate attachment, as has been described previously. The incorporation of this field plate results in higher $V_{\text{CRIT}}$ values (due to reduction in peak electric field) than those reported on non-field plated devices, but the mechanics of electrical degradation appear to be similar in both cases. It is also important to note that these devices possess a 2µm gate length, rather than the 100nm gate lengths of the devices mentioned just previously. So, the banding defect observed here extends over the entire region where the gate contact of the devices contacts the surface of the AlGaN layer. In this case, the devices were stressed on the same HP4146C testing station as the devices studied in other sections of this dissertation and were step stressed with $V_{\text{GS}} = -6\text{V}$ and $V_{\text{DS}}$ stepped from -10V to -100V in 1V/min increments.

Interestingly, a similar device which was stressed at $V_{\text{DS}} = 0\text{V}$, rather than $V_{\text{DS}} = 5\text{V}$ did not manifest banding as the critical voltage was attained. This stands in stark contrast to the
device stressed with a 15V potential applied across the source and drain electrodes, which manifested banding over a small region (10% percent of the total gate area) of the device footprint, as shown in Figure 5-3B. The device which manifested banding also manifested a much larger increase in total gate leakage current as it achieved \( V_{\text{CRIT}} \). The behavior noted here might be the general trend for banding: devices stressed with a potential applied between the source and drain electrodes of the device manifest banding while devices stressed with no potential applied between the source and drain do not manifest banding. It bears noting that both of these devices were stressed to potentials much larger than the critical voltage, but no pitting was observed in either device and the gate leakage current did not increase dramatically past \( V_{\text{CRIT}} \).

The implication of this result is that banding requires potential across the 2DEG in order to occur. It may be that the mechanism which influences banding defect formation requires the presence of free electrons within the 2DEG to aid in an electrochemical reaction which degrades the quality of the gate electrode.

The formation of a banding defect can also be induced by means of thermal annealing of a Ni-gated device. Figure 5-4B demonstrates that a banding defect, with morphology similar to that of the banding defects described previously was observed after deprocessing of a typical device annealed in a Lindberg Furnace at 500ºC for 30 minutes in an air ambient. This defect was present in devices with both 1µm gate lengths and 100nm gate lengths and the morphology of this defect was the same between both devices. In contrast to the results demonstrated previously during stressing of the 100nm gate devices, the banding defect observed on these devices only extends over the region where the gate electrode directly makes contact with the AlGaN epitaxial layer. The wider and lighter banding region which flanks the darkly banded
region on the T-gate device is absent. This seems to imply that banding induced by thermal annealing does not proceed with to the same equilibrium state, or with the same rapidity, or both as banding induced by electrostatic stressing. It stands to reason that a lighter feature which exhibits less contrast in SEM is induced by material which is a poorer insulator (and which, therefore, has a smaller workfunction and corresponding secondary electron yield), or which may be thinner, or both.

In order to understand why this difference in morphology arises, it is important to consider the field present between the gate electrode and the 2DEG of the device in the case of electrical stressing and in the case of thermal annealing. The vertical electric field present 1nm below the surface of the AlGaN in these devices was simulated using the Florida Object Oriented Device Simulator (FLOODS) and is shown both in the case of stepped stressing up to $V_{\text{CRIT}}$, and annealing at 500ºC as part of Figure 5-4C and Figure 5-4D, respectively. The device which was simulated had a gate length of 100nm and had a total length of 500nm at the top of the T-gate. The crossbar of the T-gate was raised 90nm above the surface of the AlGaN. The AlGaN layer, itself, was 15nm thick and formed on top of a 2µm GaN buffer layer. In the case of stepped stressing up to $V_{\text{CRIT}}$, potentials of $V_{DS} = 5$ V and $V_{GS} = -22$ V were applied to the electrodes of the devices. In the case of annealing at 500ºC, potentials of 0 V were applied to all electrodes.

During electrical stressing, the gate contact is placed under a high negative potential. Electric field permeates the device underneath the entire gate length, but is maximized where the gate makes direct contact with the AlGaN. In this region, the vertical electric field achieves a value of $4.3 \times 10^6$ V/cm. A lower field exists in the region of the AlGaN where the gate overhangs the surface, which is unsurprising given that the electric field between two conducting
layers is inversely proportional to the distance between those layers. The field under the edge of
the T-gate is equal to $4.8 \times 10^5$ V/cm.

In the case of thermal annealing, a null potential is applied to the gate electrode. As a
result, the only field present between the gate electrode and the 2DEG arises from the built-in
field (and associated dipole) generated by the difference in workfunction between the gate
electrode and the AlGaN layer beneath it. So, field is only present where the gate electrode
makes direct physical contact with the AlGaN epitaxial layer. The vertical field directly under
the gate in this case is significantly reduced, only achieving a value of $4.8 \times 10^5$ V/cm, and is
more-or-less confined to the region directly under the gate.

It bears noting that the field under the gate, in this case, has a value which is of the same
order of magnitude as the vertical electric field present at the furthest extent of the T-gate in the
previous simulation. If vertical electric field is the driving factor for banding, this would suggest
that the band which forms under the gate foot in an annealed device should have reacted less
than the bands which form under the raised portion of the T-gate during annealing at 500ºC. The
SEM micrographs in Figure 5-4A and Figure 5-4B suggest that this is, indeed, the case.

The implication of the differences in morphology observed between defect formation by
means of electrical stressing and defect formation by means of thermal annealing is that both
field and temperature are required to induce defect formation. Thus, banding is some sort of
electrochemical reaction, similar to phenomena in other semiconductors such as time-dependant
dielectric breakdown [120]. It is dependant upon the field and ambient temperature present
under the gate electrode of the device and it also appears to require a steady supply of free
electrons within the 2DEG.
In order to demonstrate causality between banding and the increase in gate leakage current rather than mere correlation, the $I_{GS}$-$V_{GS}$ current characteristics of the gate contact of an AlGaN / GaN HEMT were analyzed both before and after annealing at 500°C for 30 minutes. The results of this study can be seen in Figure 5-5. A slight (2X) increase in the gate leakage current is observed at large voltages after thermal annealing at 500°C. Leakage current increased to a much lesser degree in the annealed device than in the device which was electrostatically stressed, further suggesting that thermal annealing does not initiate degradation which is as severe as degradation which occurs when the vertical electric field is much larger. This fits well with observations of banding in SEM.

SPM analysis was performed on the surface of a sample which formed the observed "banding" defect after annealing at 500°C for 30 min. This SPM was performed with a Veeco Dimension 3100 SPM in tapping mode. The integral gain was set at 5.0 and proportional gain was also set at 5.0 for all testing while the photodiode amplitude setpoint was held at 350mV in order to ensure good coupling between the AFM tip and the surface. A TESP-HAR AFM tip, available from Bruker Nanosurfaces, with a 10nm tip diameter, a 40 N/m spring constant, and a 5:1 aspect ratio was used for this study. The region studied as part of SPM analysis was the channel region of the device, represented as a 512x512 pixel topographic map, as measured by AFM. This 512x512 array is representative of a 6.75µm x 6.75µm section of the channel region of the device.

As shown in Figure 5-6A, the dark band of contrast which appeared in SEM and which matched closely with the dimensions of the 1µm Ni-gate is still visible during AFM analysis. A 3.5µm wide region of the channel was integrated in order to analyze the thickness of this banded region. According to the averaged linescan resulting from this analysis, shown in Figure 5-6B,
this dark band of contrast corresponds with a raised region on the surface of the AlGaN epitaxial layer possessing a height of approximately 1.0 nm in height. It is interesting that this banded region appears to possess peaked edges, which correspond to the peaked nature of the vertical electric field present at the edges of the gate contacts of the simulated devices.

The roughness of this region differs from the roughness of the surrounding deprocessed AlGaN. The RMS roughness of the channel region was determined by means of a second measurement with AFM, this time of a 1.5 x 1.5 µm section inside and outside of the "banded" defect region within the channel. All other settings, except for the dimensions of the scanned area, were maintained between the two measurements. Analysis of the Root Mean Square (RMS) Surface Roughness was performed on both regions. The channel region of the device, in absence of banding, was found to have a surface roughness approximately equal to 3.7Å with a standard deviation of 1.6Å, while the surface roughness of the banded region was determined to be approximately equal to 3.4Å with a standard deviation of 1.1Å. This would suggest that the material being formed as a result of thermal annealing is likely the product of a reaction at the epitaxial interface between the AlGaN and the Ni. It is interesting to note that the integrated AFM linescan of the band present under the gate manifests a pair of "peaks" where the edges of the contact were present. This agrees with simulation of the vertical electric field, where a pair of peaks was observed at the edges of the gate contact.

This AFM analysis does not explain why this banded region is present in annealed and stressed samples in the first place, however. In order to explain the cause of band formation as well as the mechanism of its formation, information about the composition of this defective region is required. Arguably, the best means of extracting this information is direct observation
of the banded defect with TEM. To this end, cross-sectional TEM analysis was performed on two separate samples the interface between the AlGaN and the Ni-gate of a HEMT device.

A control sample was formed from a device which was received as-formed from fab processing and a second sample was formed from a device which was annealed at 500ºC for 30 min in order to induce the formation of a banding defect. This device was marked with fiduciaries which set off the gate region of the device, deprocessed using the etch chemistry described previously and coated with thermally evaporated carbon to ensure that the gallium ion beam of the FIB did not interact with the banding defect present on the surface of the AlGaN.

As shown in Figure 5-7A, HAADF-STEM analysis of the control device, performed with a JEOL-ARM200F, demonstrates the convenient presence of an interfacial layer between the AlGaN surface of the HEMT device and the nickel gate of this device. The thickness of this interfacial layer is approximately 1.5nm prior to annealing and approximately 2nm post-anneal. This increase in the observed thickness of the interfacial region indicates that it may have reacted under the gate electrode of the device. The difference in observed thickness between as-fabricated and annealed devices could be due to consumption of the AlGaN, which appears to be occurring after annealing, as shown in Figure 7-B. It could also be due to stochastic variations in the thickness of the interfacial layer. The author notes that the band height extracted from many linescanes of the banded region are much larger than the "averaged" linescan presented in Figure 5-6B, often in excess of 2 nm.

Stressing by means of thermal annealing or electrostatic biasing appears to induce some sort of chemical change within this interfacial layer which makes it much less pervious to the etch chemistries used for sample preparation in plan-view SEM analysis and which induces
higher leakage currents within the gate contact of the HEMT. This interfacial layer remains after deprocessing, as shown in Figure 5-7B.

In order to analyze the order present with this interfacial layer, HRTEM was employed, as shown in Figure 5-8. Analysis of the Fast Fourier Transform (FFT) of the interfacial layer (processed with Image J® software, which is open source and available at http://rsbweb.nih.gov/ij/) suggests that this region is amorphous in nature both before and after the formation of the "banded' defect present under the gate. The FFT of this region is dominated by a bright central disk, indicative of the nearest neighbor distance associated with the amorphous region. The bright array of spots associated with a single crystal sample are present within this FFT, but are dim and match closely with the positioning of spots in the FFT of the underlying AlGaN layer. This indicates that the observed dim diffraction spots in the FFT of the interface do not arise unique crystalline phases within this layer, but are more likely induced by thickness variation within the cross-sectioned sample.

In order to determine the chemical composition of the interfacial layer, EELS and EDS were performed in conjunction with HAADF-STEM of the interfacial region. As shown in Figure 5-9A, EELS analysis reveals that this interfacial layer is oxygen-rich in comparison to the surrounding AlGaN regions and that this remains the case both before and after annealing and deprocessing of the device structure. The EELS signal arising from N is present along with Ga this interfacial layer. However, it is noted that the brightness of the N and Ga signal is not as substantial as the brightness in the AlGaN layer. Upon annealing, the signal arising from N in this layer is greatly diminished, as is shown in Figure 5-9B. The signal arising from Ga is also diminished and diffuse. The signal from O also appears to be diffuse. It appears that, after
annealing, the native oxide which is present on the surface of the AlGaN reforms, rejecting N, and grows. Ga outdiffusion also appears to occur.

The presence of an oxide interface between the AlGaN and Ni-gate fits well with previous studies (via Atom Probe Tomography) of the interfacial region of the Ni-gated HEMT gate stack. These studies suggested that a native oxide forms between these two regions if a sample is taken out of a high-vacuum environment in between deposition of the AlGaN epitaxial layer of the HEMT via MBE and the deposition of the nickel gate of the device [113]. The native oxide present within the commerial device analyzed as part of the APT studies was thinner than the native oxide present in this study, however, and it is difficult to draw any conclusions from this previous research (other than that a native oxide interface should, likely, exist between the AlGaN and nickel) because of the different processing conditions inherent in manufacturing of these two different HEMT architectures.

In order to ascertain more information about the chemical structure of this native oxide, EDS was performed in conjunction with HAADF-STEM. EDS was chosen as a complementary technique to EELS because of the large overlap between the Gallium signal in the EELS spectra and the peaks associated with Aluminum and Nickel. The superposition of these two peaks made even qualitative analysis of the atomic concentrations of aluminum and nickel within the interfacial layer much more difficult.

Figure 5-10 shows the EDS spectra associated with a STEM linescan across the interfacial layer in samples formed both before and after thermal annealing at 500ºC for 30 min of the HEMT device structure. The regions over which these linescans formed are depicted in the inset figures contained within each linescan. Special attention should be paid to the x-ray
peak corresponding to the Kα line of aluminum, which varies considerably between the control and the annealed device, and its relation to the Ga Kα, Ni Kα, O Kα, and Au Lα lines.

Three general regions of interest are contained within each of these EDS linescans. The deepest region of interest is the AlGaN region itself, which is dominated by EDS signals from the Ga Kα and Al Kα lines. The shallowest region of interest in this linescan is the gate electrode, where the Ni Kα and Au Lα lines dominate. It bears noting that the Ni Kα peak dominates the EDS signal in this region in the control sample while the Au Lα and Ni Kα peaks both demonstrate large signals in the EDS spectrum in the annealed sample. This is due to intermixing between the nickel liner layer of the gate electrode, which is approximately 20nm thick prior to annealing, with the gold layer that acts as the contact pad for probing of the device. Between these two regions is the interfacial oxide present between the AlGaN and the gate metal, which coincides with the slight increase in the EDS signal arising from the O Kα peak. This interface is formed when vacuum is broken after the deposition of the AlGaN epitaxial layers, prior to the formation of the gate electrode via lithography and metal deposition and it is the changes which occur in this interface during annealing which are of interest in this study.

The EDS linescan shown in Figure 5-10B demonstrates that, upon annealing at 500ºC for 30min, a peak in the Al Kα EDS signal forms within this interfacial layer. This peak is accompanied by a local depression in the same Al Kα EDS signal just prior to the interfacial layer. This result is consistent over a range of different regions and is consistent from one annealed sample to another. What appears to be occurring is the segregation of the aluminum present within the AlGaN into the native oxide layer formed between the AlGaN epitaxial layer and the gate electrode of the device, causing the increase in the Al Kα signal at the interfacial oxide. As aluminum segregates into the interfacial oxide from the AlGaN just below, a region of
aluminum-deficient AlGaN forms, leading to the local depresssion in the Al Kα signal arising from the AlGaN epitaxial layer just below the native oxide interface.

This segregation during annealing may be due to the restructuring of the interfacial oxide into the most thermodynamically stable phase. It is well known that gallium oxides do not possess the characteristic thermal stability enjoyed by aluminum oxides at elevated temperatures. In fact, Holzworth and coworkers demonstrated previously on a commercial sample that it is common for the oxide present at this interface to be formed from alumina rather than some gallium oxide [113]. It is possible that, upon annealing or electrostatic stressing, the interfacial oxide undergoes a phase change, converting from some disordered oxidized state to some form of an aluminum oxide. This thermally stable aluminum oxide would be more resistant to BOE etching and could give rise to the feature observed in SEM and AFM.

The formation of this alumina interfacial layer would also change the leakage observed at the gate contact, as the dipole formation associated with fermi-level pinning of the semiconductor at the gate electrode is highly dependant upon interfacial quality. It bears noting that the thickness associated with maximum conduction of an Al contact to Si with an Al₂O₃ interfacial layer is approximately 1.5nm, which fits well with observations made on these devices [121].

5.3 Defect Formation Beyond the Critical Voltage

The initial study described at the start of this chapter is revisited in Figure 5-11, which demonstrates the quality of the surface of the AlGaN epitaxial layer as a device is step stressed at -1V/min well beyond the critical voltage. As shown in this figure, the banding characteristic to device degradation at \(V_{\text{CRIT}}\) is no longer visible.

It is understood that further electrostatic stressing above the critical voltage induces substantial reactivity and mobility of nickel atoms present within the liner layer of the gate.
electrode. This layer may getter the oxygen from the interface, effectively consuming it, causing the strange "disappearance" of the banding defect. It is also possible that, in cases where the voltage is driven up above \( V_{\text{CRIT}} \) rapidly, banding only forms in very isolated regions of the gate contact. The increase in leakage current density appears to be substantial when banding occurs via electrostatic stressing. After all, the total gate current in the device increased by two orders of magnitude as a result of banding over 66% of the surface of the device. This implies that the power dissipated through the interfacial oxide per unit area as a result of banding should increase by approximately two orders of magnitude, which is a substantial increase.

It seems plausible that this increase in gate current due to banding could result in substantial joule heating in isolated regions of the gate contact. The temperature rise as a result of joule heating could enhance the reaction which forms the pitting defect which is so commonly observed at voltages well above \( V_{\text{CRIT}} \). This could even be the case in situations where the voltage applied between the source and drain is zero, thanks to trap assisted tunneling of electrons from the 2DEG or local conduction through threading dislocations.

The banding defect is replaced by the well-documented pitting defect observed by various authors in Ni-gated devices. This pitting defect covers approximately 5.5% of the total aerial density of the 100nm gate studied as part of the experiment. While the pitting defect has been observed previously by various experimentalists, the nature of its formation and how this defect affects the current characteristics of the HEMT is not well-understood for pitting defects induced in a Ni-gated device. To that end, an additional experiment was performed to ascertain the dependance of gate leakage current on the formation of pitting defects under the gate electrode.
As before, devices were stressed in an air ambient using a Tektronix 370A curve tracer and an HP 4156 parameter analyzer. The devices were placed in a deeply pinched-off mode (with gate voltage $V_{GS} = -8\,\text{V}$) to ensure minimal current flow between the source and drain electrodes as well as to maximize the field present between the gate and the 2DEG. The voltage between the gate electrode and the drain, $V_{DG}$, was increased at a rate of $-1\,\text{V/min}$ from a value of 5V to as much as 70V or until the device achieved current compliance, which varied from 500$\mu\text{A}$ to 1mA, depending on the device under stress. The gate leakage current during stress as well as in "off" mode, with $V_{GS} = -1\,\text{V}$, was measured during the experiment. As before, this measurement occurred over a time frame approximately equal to 1ms, meaning that the relative duty cycle associated with stressing was, effectively, 100%.

The leakage at the end of this stressing period for seven separate devices was correlated with the density of pitting defects which were observed after standard deprocessing utilizing FeCN/KI combined with HF. This measurement was performed, as before, on an FEI SEM in UHR mode utilizing a TLD in secondary electron mode and the image analysis associated with estimating the aerial density of pitting defects was performed with Image J ©. This aerial density was normalized to the area of each device under analysis in order to yield a density in terms of the percent of the gate area which reacted.

The result is shown in Figure 5-12. A direct correlation can be observed between the leakage current through the gate electrode during off-mode stressing and the observed aerial density of pitting defects. This is true for devices with large gate lengths and for devices with aggressively scaled gate lengths. However, it bears noting that defects which form under larger gates appear to induce larger leakage densities through these pitting defects than those which form in aggressively scaled devices.
A metallic Ni phase would be a more effective conductor than an oxidized Ni phase. Thus, the magnitude of the leakage current density should vary depending upon whether the defects present under the gate are comprised of metallic or oxidized Ni. In devices with 100 nm gate lengths, the diffusion length required for O\textsubscript{2} to permeate the entire gate length after diffusing through the SiN\textsubscript{x} passivation layer is as little as ~100 nm, while the diffusion length required to permeate the entire gate length of a device with 1.0 μm gate length is ~550 nm. Therefore, for a device with 100 nm gate length, O\textsubscript{2} may diffuse enough during stressing such that the majority of defects generated are oxidized Ni-based rather than metallic Ni-based, resulting in lower leakage current densities for those devices over the 1μm devices.
Figure 5-1 A plot of gate current degradation and associated defect formation prior to $V_{\text{CRIT}}$. A) Gate current is compared to the source and drain currents to generate leakage current components passing through the gate electrode from the two contacting electrodes. B) An exemplary scanning electron micrograph of the deprocessed surface of this stressed device. No evidence exists of defect formation during this regime of stressing.
Figure 5-2  A plot of the gate degradation up to a voltage approximating $V_{\text{CRIT}}$. A) At $V_{\text{CRIT}}$, the leakage current through the gate electrode both during stressing and in off-mode, increases by two orders of magnitude. B) An exemplary scanning electron micrograph of the deprocessed surface of this stressed device. A pair of inset dark bands in the same aerial position as the t-gate are observed.
Figure 5-3  A plot of the degradation of a circular HEMT, with the same "banding" defect.  A) Induced gate leakage current resulting from stepped stressing, at a rate of $V_{GS}=-1V/min$ and starting at $V_{GS}=-10V$ of a circular HEMT with $V_{DS}=15V$, indicates that $V_{CRIT}$ is achieved and that no additional leakage current is induced past $V_{CRIT}$.  B) A scanning electron micrograph formed by an FEI NovaSEM operating with the TLD detector and the immersion lens, reveals that this stepped stressing resulted in the formation of a band-like defect under 10% of the total gate area.
Figure 5-4  Morphological differences in banding induced by electrostatic stress and annealing. 
A) Stepped stressing results in the formation of a pair of inset dark bands on the 
AlGaN surface where the nickel gate was once present.  B) A similar single band 
defect can be observed on the AlGaN surface in a device which was annealed in a 
Lindberg Furnace at 500ºC for 30 min.
Figure 5-4 (Continued) C) FLOOPS was used to simulate the vertical electric field present 1nm below the AlGaN surface in the stressed device at $V_{GS} = -22 \text{ V}$ and $V_{GS} = 5 \text{ V}$. D) It was also used to simulate the vertical electric field present in the device with all electrodes of the device set to a potential of 0 V when that device is held at a temperature of 500ºC.
Figure 5-5 The gate leakage current before and after annealing at 500°C for 30 min. A small increase in gate leakage current is observed, which is commensurate with themally-induced banding under the gate electrode of the device.
Figure 5-6  SPM of an AlGaN/GaN HEMT annealed at 500°C for 30 min.  A) A raised band of roughness, comparable to the band observed in SEM, was detected.  B) An average linescan of this band over 5µm of the device width reveals that it is approximately 1.0 nm in thickness.
Figure 5-7  HAADF-TEM of the interface before and after annealing. A) Analysis of lamella formed from an as-recieved device indicates the presence of a thin interfacial region which is amorphous in nature and ~1.5nm. B) Cross-sectional BF-TEM of a lamella formed from a device which was annealed at 500°C for 30min and then deprocessed indicates that this interfacial region is preserved, if some intermixing does appear to occur.
Figure 5-8  HRTEM and FFTs of the Ni/AlGaN interface before and after annealing. A) Analysis of lamella formed from an as-received device indicates the presence of a thin interfacial region which is amorphous in nature and roughly equal in thickness (~1.5nm). B) Cross-sectional BF-TEM of a lamella formed from a device which was annealed at 500ºC for 30min and then deprocessed indicates that this interfacial region is preserved, if some intermixing does appear to occur.
Figure 5-9  HAADF-STEM and EELS of the Ni/AlGaN interface before and after annealing.  A) Analysis of a lamella formed from an as-received sample indicates that the thin interfacial region separating the AlGaN and Ni gate is comprised of oxygen, indicating that a native oxide may have formed.  B) Cross-sectional DF-STEM, in conjunction with EELS, of a lamella formed from a sample annealed at 500ºC for 30min and deprocessed, indicates that this layer remains oxygen rich, if somewhat more diffuse, after annealing.
Figure 5-10  An EDS linescan of the Ni/AlGaN interface before and after annealing. A) Analysis of the linescan from a sample as-fabricated indicates that the EDS signals arising from Al and Ga are constant up through the interfacial region until they hit the Ni of the gate electrode. B) An EDS linescan of this same interfacial region, after annealing at 500°C for 30s and deprocessing in an etch solution, indicates a segregation of aluminum into this interfacial layer. The EDS signal arising from the O Kα line also appears to be enhanced.
Figure 5-11 Stepped stressing of a device to voltages well in excess of $V_{\text{CRIT}}$. This stressing eliminates the banding observed earlier at the critical voltage. Pitting of the AlGaN layer, as was observed previously by Holzworth et al., is observed in the channel region previously occupied by the gate electrode.
Figure 5-12  Percentage of gate contact area consumed by under-gate defects. This metric is a function of gate leakage current density (prior to catastrophic failure) for both nm- and µm-scale devices.
CHAPTER 6  
CONCLUSIONS

Because of the highly stochastic nature of defect formation in AlGaN/GaN HEMTs, techniques which only sample small volumes samples from a much larger device (such as TEM) are poorly suited in analytical studies. Deprocessing and analysis utilizing top-down SEM or AFM is often preferable to TEM for many studies. A deprocessing scheme was devised which utilizes a 15 minute exposure in BOE followed by a 28hr exposure in a mixture of FeCN and Ki, otherwise known as TFAC. After degreasing via ultrasonic cleaning in a 1:1 mixture of n-heptane and acetone for 2hr, a 2hr ultrasonication in methanol and a 30 minute exposure in DI water, a clean AlGaN surface can be imaged utilizing UHR-SEM. This deprocessing technique has been used with great effect in studies of defects in as-formed samples as well as in samples which were stressed electrostatically.

Investigations of AlGaN/GaN HEMTs formed with Ti/Al/Ni/Au ohmic contacts via annealing at 850°C for 30s have revealed the formation of TiN inclusions which form at the AlGaN interface. These inclusions generally form in the vicinity of a threading dislocation, which is consistent with a model which involves inclusion formation via metal diffusion down a dislocation core. A layer of Al-rich material was also observed at the interface between the TiN and the AlGaN surrounding it. Analysis with HAADF-STEM and EDS indicate that the TiN inclusion is devoid of Ga. The Ga which was present in the epitaxial layer in place of the TiN may be insoluable in these metal inclusions and is pushed out into the remainder of the AlGaN in a front around the metal inclusion.

The inclusions themselves appear to have a faceted morphology, which fits with the substantial energetic differences between different wurtzite GaN surfaces. XTEM demonstrates that, as these inclusions form and push Ga out into the surrounding AlGaN layer, cracks nucleate
at the corners of their faceted sidewalls along the [112-0], or prism directions. It also demonstrates that the cracks themselves attain depths approximately equal to the AlGaN thickness. Nanocracks grow differently after their nucleation depending upon their location in relation to the edges of the source and drain contacts.

Cracks which nucleate in the interior of the ohmic contact encounter substantial compressive stresses due to the presence of a substantial aerial density of metal inclusions. Because of this, cracks which nucleate in the ohmic contact, due to tensile hoop stress brought on by compressive stresses oriented radially around the metal inclusion, are short – with the maximum of the distribution of their crack lengths being approximately equal to 40nm-60nm. Cracks which nucleate at the edges of the ohmic contact propagate into a region of the AlGaN/GaN epitaxial layer which is not under substantial compressive stress, as no metal inclusions are present. The cracks encounter the tensile epitaxial stresses of the AlGaN layer and grow to much longer lengths, resulting in a distribution of cracks lengths with a maximum at 140nm and with a much larger variance than the distribution associated with cracks present in the interior of the ohmic contacts.

It should be noted that channel cracks at the long end of the distribution require a tensile stress for growth on the order of the those associated with the extreme voltages attained during electrostatic off-mode stressing. This is of interest given results from electrostatic stressing of an AlGaN/GaN HEMT stressed at $V_{DS} = 15V$ and from $V_{GS}=-10V$ to $V_{GS}=-42V$ at a rate of -1V/min. Deprocessing followed by Plan-View SEM analysis revealed that the lengths of nanocracks present in the channel of this device were much longer than cracks observed in as-formed devices, suggesting that stressing may have induced further crack growth. The degradation of $I_{DS}$ and $V_{GS}$ in the device might be commensurate with this phenomenon, as
cracks in AlGaN/GaN epitaxial structures can induce a shorting of the gate contact to the source or drain if they extend under the gate contact itself.

This result illustrates a fundamental flaw of ohmic contact formation via TiN metal inclusions. It stands to reason that, in order to further reduce the contact resistance of an alloyed metallurgical contact to AlGaN/GaN, the total area over which the TiN directly contacts the 2DEG must be maximized via wholesale reaction of Ti with AlGaN. It also stands to reason that, as the density of inclusions increases, so too will the density of nucleated cracks within the ohmic contacts and within the channel region. Cracks extending into the channel region will degrade the device by means of gate shorting. An alternative method of contact formation or some adjustment of architecture to trap cracks which form on TiN inclusions may be required in order to further improve the reliability of devices under stress.

Analysis of devices stressed from $V_{GS} = -10V$ up to values of $V_{GS} < V_{CRIT}$, $V_{GS} = V_{CRIT}$, $V_{GS} > V_{CRIT}$ at a rate of -1V/min with $V_{DS} = 5V$ revealed the relationship between the increase in $I_G$ at $V_{CRIT}$ and a structural change in the AlGaN surface of a HEMT stressed with a bias applied between the source and drain electrodes. This new defect was a "band" of dark contrast which was observed under a 100nm Lg gate electrode in UHR-SEM utilizing a TLD. A similar dark band of contrast can be induced by annealing a device at 500°C for 30min. The differences in morphology of these two bands can be explained by an field-induced reaction mechanism where applied electric field impacts the activation energy of a reaction, similar to time dependant dielectric breakdown.

It should be noted that this defect only appears in devices which were stressed with a potential applied across the source and drain. It may be that a steady supply of electrons from the 2DEG to the interface of the device is necessary for the formation of this banding defect.
The $I_G-V_{GS}$ characteristics of an AlGaN/GaN HEMT after annealing at 500°C for 30min reveals an increase in the reverse-biased gate leakage current. The increase is not as dramatic as the increase which was induced in the AlGaN/GaN HEMT after electrostatic stressing, but it should be noted that the contrast of the band observed in the annealed sample is also less pronounced than the sample which was stressed electrostatically. Presumably, thermal annealing with only the built-in field of the schottky contact to reduce the activation energy associated with the reaction does not induce as high a rate of reaction as stressing with a high applied field.

AFM analysis of the banded surface of an AlGaN/GaN HEMT with a 1µm gate length reveals that the banding observed in SEM corresponds to a raised feature with a thickness of approximately 1.75nm. This defect has approximately the same surface roughness as the channel, suggesting that it may be formed by some reaction at the interface between the epitaxially deposited AlGaN and the gate contact. Interestingly, the thickness of the band as it was observed in AFM matches the thickness of the native oxide layer formed at the interface of the AlGaN prior to gate metal deposition. STEM combined with EELS reveals that this oxide layer is amorphous in nature and is present both before and after annealing at 500°C. EDS analysis of the interface revealed that, upon annealing, aluminum appears to segregate into the native oxide layer from the AlGaN.

As stressing proceeds past the critical voltage, the band defect is no-longer present. This strange case of disappearance may be due to the tendency of nickel present within the gate to react with oxygen present in the ambient and in the device. Given that pitting defect formation appears to occur after $V_{CRIT}$ in this stressing regime, it may be that banding is a process with a lower inherent activation energy than the reaction by which Ni reacts with O and pits the surface.
Analysis of other devices stressed above the critical voltage via deprocessing and top-down SEM indicate that the pitting reaction has a direct impact on the reverse biased gate leakage observed in an AlGaN/GaN HEMT after stressing well beyond $V_{\text{CRIT}}$. The aerial density of pits observed in SEM appears to influence directly the reverse biased leakage current, which fits with pitting forming an ohmic contact to the 2DEG and effectively shorting out the gate contact of the device.

The formation of banding and pitting based defects has a substantial impact on the overall reliability of the gate contact of AlGaN/GaN HEMT technologies. It may be that these reliability issues cannot be engineered out without substantial changes to device architecture. Introduction of a thermally stable oxide layer as well as a change to a more inert gate metal, such as Pt, might alleviate banding and pitting by fundamentally changing the nature of the contact of the device as well as its reactivity. A MOS architecture which also makes use of a field plate to reduce the peak electric field under the gate contact seems like a very promising candidate for eliminating the reliability issues inherent to the contacting of a Ni schottky gate directly to AlGaN.
LIST OF REFERENCES


BIOGRAPHICAL SKETCH

Patrick Whiting was born in Chicago, Illinois in the United States of America in 1986 and he grew into adulthood in East Bloomfield, New York – a small town outside of the city of Rochester, nestled snugly in the Finger Lakes region of the state. He graduated from high school in 2004, was the salutatorian of his class, and was voted "most likely to succeed". His interest in electronics and the growing field of nanotechnology brought him to Rochester Institute of Technology, located in West Henrietta, New York.

He began his career as a researcher in his freshman year of school, studying fresnel microlens arrays, sputtered thin film dielectrics for MRI, high power silicon electronics and single-crystal thin film transistors. In the winter of 2009, he graduated Magna cum Laude with his BS in microelectronic engineering and his MS in materials science and engineering. Faced with the horrible economic climate of a worldwide recession and with a ravenous hunger for more knowledge, Patrick decided that he would pursue a Ph.D in materials science and engineering from the University of Florida under Dr. Kevin S. Jones. He studied AlGaN/GaN High Electron Mobility Transistors, a new technology which was becoming popular for a variety of applications, including the infamous "naked scanners" which can be found at most domestic airports.

Patrick will graduate in December 2013 and move on to work for Intel Corporation in Hillsboro, Oregon where he will be an engineer working on process development for next-generation microprocessors. Patrick expects that, when he isn't working, he will be found in the environs around the city of Portland, usually writing some crazy science fiction story or other.