DEVELOPMENT OF HIGH TEMPERATURE STABLE OHMIC AND SCHOTTKY CONTACTS ON N-GaN

By

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A DISSERTATION PRESENTED TO THE GRADUATE SCHOOL OF THE UNIVERSITY OF FLORIDA IN PARTIAL FULFILLMENT OF THE REQUIREMENTS FOR THE DEGREE OF DOCTOR OF PHILOSOPHY UNIVERSITY OF FLORIDA 2007
To my family
ACKNOWLEDGMENTS

I would like to thank my advisor, Prof. Stephen J. Pearton, the most important person throughout my graduate studies, for all the opportunities, guidance, motivation, and support. I would also like to thank my committee members, Prof. Cammy R. Abernathy, Prof. David P. Norton, Prof. Fan Ren, and Prof. Rajiv Singh, for their time, expertise, and evaluation.

Prof Pearton has been my mentor in true sense. He provided me numerous opportunities to give presentations, encouraged logical, problem solving thinking and gave me confidence whenever I needed it. Prof. Pearton helped me develop wide ranging skills in semiconductor processing area for which I am really thankful to him. I can not thank him enough. I thank Prof. Ren for providing me with useful comments and directions in order to improve my research work. I am grateful to them for their advice that has helped me grow professionally.

I would like to thank group members of Prof. Pearton, Prof. Ren, Prof. Abernathy, Prof Norton, and Prof Singh research groups’, Kwang Baik, Kelly Ip, Lars Voss, Jon Wright, Wantae Lim, Rishabh Mehandru, Soohwan Jang, Byong Kang, Hung-Ta-Wang, Travis. J Anderson, J.J Chen, Luc Stafford, Brent Gila, Seemant Rawal, Karthik Ramani, Mark Hlad and countless others for their assistance and friendship and who have made graduate school enjoyable, learning experience. Especially, I am grateful to have as a very good friend, Kao-Chil-Joe, who was a visiting scholar form Taiwan and I thank him for helping me during my first year as graduate student. I also like to thank University of Florida Nano-Fabrication Lab staff, Ivan and Bill as I enjoyed working with them. I would also like to thank Paula, our group secretary for all the support she has given me.

I also want to thank all my friends of school and college with whom I share special memories forever in my life. I also want to thank my ex-roommates who made graduate life an enjoyable experience, especially Shiv for having a lot of interesting and intellectual debates.
Most importantly, I express my deepest gratitude to my family especially to my parents (my father, Prem Kumar Khanna and mother, Madhu Khanna) whose love and sacrifice for me is beyond anything I will ever understand and my brother (Bhaskar) and sisters (Ela and Rashmi) for loving and supporting me unconditionally. I thank my family for making me who I am today.
TABLE OF CONTENTS

ACKNOWLEDGMENTS ........................................................................................................4

LIST OF TABLES ..................................................................................................................9

LIST OF FIGURES ..............................................................................................................10

ABSTRACT ..........................................................................................................................15

CHAPTER

1 INTRODUCTION .............................................................................................................18

2 BACKGROUND AND LITERATURE REVIEW ..............................................................23
  Properties of GaN ..............................................................................................................23
    Overview .........................................................................................................................23
    Crystal Structure and Basic Properties ....................................................................23
    Metal Contact ................................................................................................................25
      Schottky contact ........................................................................................................26
      Ohmic contact ...........................................................................................................29
      Thermal stability ......................................................................................................32
  Common Processing Techniques ..................................................................................33
    Dry Plasma Etching .....................................................................................................33
    Ion Implantation .........................................................................................................36
    Rapid Thermal Annealing .........................................................................................37
  Characterization Techniques .......................................................................................37
    Atomic Force Microscopy .........................................................................................37
    Auger Electron Spectroscopy ....................................................................................38
    X-ray Photoelectron Spectroscopy ..........................................................................38
    Electrical Measurements ............................................................................................38
    Photoluminescence (PL) ...........................................................................................39
    Rutherford Backscattering Spectrometry/Channeling ............................................39
    Scanning Electron Microscopy ................................................................................40
    Secondary Ion Mass Spectrometry ..........................................................................40
    Stylus Profilometry .....................................................................................................41

3 TUNGSTEN AND ZIRCONIUM BORIDE BASED OHMIC CONTACTS TO N-GaN ....53
  Introduction ....................................................................................................................53
  Experimental ..................................................................................................................54
  Results and Discussion .................................................................................................55
    Tungsten Boride Based Ohmic Contact ....................................................................55
    Zirconium Boride Based Ohmic Contacts ..................................................................58
  Summary and Conclusions ...........................................................................................60
4 COMPARISON OF ELECTRICAL AND RELIABILITY PERFORMANCE OF TiB₂, CrB₂ AND W₂B₅ BASED OHMIC CONTACTS ON N-GaN ..............................................73

Introduction........................................................................................................73
Experimental........................................................................................................74
Results and Discussion .......................................................................................75
Summary and Conclusions ...............................................................................77

5 ZrB₂ AND W₂B SCHOTTKY DIODE CONTACTS ON N-GaN..........................87

Introduction ..........................................................................................................87
Experimental.........................................................................................................88
Results and Discussion .......................................................................................90
  W₂B Based Rectifying Contacts ......................................................................90
  ZrB₂ Based Rectifying Contacts .....................................................................91
Summary and Conclusions ...............................................................................93

6 ANNEALING TEMPERATURE DEPENDENCE OF TiB₂ W₂B₅ AND CrB₂ SCHOTTKY BARRIER CONTACTS ON N-GaN .................................................106

Introduction .........................................................................................................106
Experimental.......................................................................................................107
Results and Discussion .......................................................................................109
  TiB₂ Based Schottky Contact .........................................................................109
  W₂B₅ Based Schottky Contact .......................................................................111
  CrB₂ Based Schottky Contact .......................................................................113
Summary and Conclusions ...............................................................................115

7 IMPROVED LONG-TERM THERMAL STABILITY AT 350 °C OF TiB₂ BASED OHMIC CONTACTS ON AlGaN/GaN HIGH ELECTRON MOBILITY .................135

Introduction .........................................................................................................135
Experimental.......................................................................................................136
Results and Discussion .......................................................................................137
Summary and Conclusions ...............................................................................139

8 Ir BASED SCHOTTKY AND OHMIC CONTACTS ON N-GaN........................148

Introduction .........................................................................................................148
Experimental.......................................................................................................149
Results and Discussion .......................................................................................151
  Schottky Contacts ..........................................................................................151
  Ohmic Contacts .............................................................................................152
Summary and Conclusions ...............................................................................153

9 CONCLUSIONS ................................................................................................162

LIST OF REFERENCES ..........................................................................................167
<table>
<thead>
<tr>
<th>Table</th>
<th>Description</th>
<th>Page</th>
</tr>
</thead>
<tbody>
<tr>
<td>2-1</td>
<td>Electrical properties of Si, GaAs and GaN.</td>
<td>41</td>
</tr>
<tr>
<td>2-2</td>
<td>The physical parameters in different semiconductor materials</td>
<td>42</td>
</tr>
<tr>
<td>2-3</td>
<td>Ionization energy of impurities for wurtzite GaN.</td>
<td>42</td>
</tr>
<tr>
<td>2-4</td>
<td>Basic physical properties of GaN.</td>
<td>43</td>
</tr>
<tr>
<td>2-5</td>
<td>Metal work function and ideal barrier heights for GaN (electron affinity: 4.1 eV)</td>
<td>43</td>
</tr>
<tr>
<td>3-1</td>
<td>Near-surface composition of contact stack determined by AES measurements for ZrB₂ ohmic contact</td>
<td>60</td>
</tr>
<tr>
<td>4-1</td>
<td>Selected properties of potential boride contacts on GaN.</td>
<td>78</td>
</tr>
<tr>
<td>5-1</td>
<td>Near-surface composition data obtained from AES measurements.</td>
<td>94</td>
</tr>
<tr>
<td>6-1</td>
<td>Concentration of elements detected on the as-received surfaces of TiB₂ based schottky contacts (in Atom %†)</td>
<td>116</td>
</tr>
<tr>
<td>6-2</td>
<td>Concentration of elements detected on the as-received surfaces of W₂B₅ based schottky contacts in Atom %†)</td>
<td>116</td>
</tr>
<tr>
<td>6-3</td>
<td>Concentration of elements detected on the as-received surfaces of CrB₂ based schottky contacts (in Atom %†)</td>
<td>116</td>
</tr>
</tbody>
</table>
## LIST OF FIGURES

<table>
<thead>
<tr>
<th>Figure</th>
<th>Description</th>
<th>Page</th>
</tr>
</thead>
<tbody>
<tr>
<td>2-1</td>
<td>Crystal structure of wurtzite GaN.</td>
<td>44</td>
</tr>
<tr>
<td>2-2</td>
<td>The III-V compound semiconductor tree.</td>
<td>44</td>
</tr>
<tr>
<td>2-3</td>
<td>Structure of a HEMT</td>
<td>45</td>
</tr>
<tr>
<td>2-4</td>
<td>Previous study of schottky contacts A) Index of interface behavior S as a function of the electronegativity difference of the semiconductors.</td>
<td>46</td>
</tr>
<tr>
<td>2-5</td>
<td>Lithography pattern for Schottky diode</td>
<td>47</td>
</tr>
<tr>
<td>2-6</td>
<td>Lithography pattern for linear TLM A) TLM pads. B) Plot for measurement.</td>
<td>47</td>
</tr>
<tr>
<td>2-7</td>
<td>An ICP reactor.</td>
<td>48</td>
</tr>
<tr>
<td>2-8</td>
<td>Electric and magnetic fields inside the reactor.</td>
<td>48</td>
</tr>
<tr>
<td>2-9</td>
<td>Chemical etching process. A) Generation of reactive species. B) Diffusion of reactive neutrals to surface.</td>
<td>49</td>
</tr>
<tr>
<td>2-10</td>
<td>Physical etching process. A) Generation of reactive species. B) Acceleration of ions to the surface. C) Ions bombard surface.</td>
<td>49</td>
</tr>
<tr>
<td>2-11</td>
<td>Combination of chemical and physical etching process A) Generation of reactive species.</td>
<td>50</td>
</tr>
<tr>
<td>2-12</td>
<td>Ion implantation system.</td>
<td>50</td>
</tr>
<tr>
<td>2-13</td>
<td>Simplified principle of AFM.</td>
<td>51</td>
</tr>
<tr>
<td>2-14</td>
<td>Auger Process. A) An isolated atom. B) Inner core level electron dislodged, leaving behind a vacancy. C) An outer level electron fills the vacancy.</td>
<td>52</td>
</tr>
<tr>
<td>3-1</td>
<td>Specific contact resistivity versus anneal temperature for Ti/Al/W₂B/Ti/Au on n-GaN.</td>
<td>61</td>
</tr>
<tr>
<td>3-2</td>
<td>Measurement as a function of annealing temperature Ti/Al/W₂B/Ti/Au on n-GaN. A) Transfer resistance. B) Sheet resistance.</td>
<td>62</td>
</tr>
<tr>
<td>3-3</td>
<td>Specific contact resistance versus measurement temperature for Ti/Al/W₂B/Ti/Au on n-GaN annealed at 800 °C.</td>
<td>63</td>
</tr>
<tr>
<td>3-4</td>
<td>Secondary electron images of the Ti/Al/W₂B/Ti/Au contacts on n-GaN. A) As-deposited. B) Annealed at 500 °C. C) Annealed at 800 °C. D) Annealed at 1000 °C.</td>
<td>64</td>
</tr>
</tbody>
</table>
AES depth profiles of the Ti/Al/W₂B/Ti/Au on n-GaN. A) As-deposited. B) Annealed at 500 °C. C) Annealed at 800 °C. D) Annealed at 1000 °C.

Contact resistance of the Ti/Al/W₂B/Ti/Au on n-GaN, initially annealed at 800 °C, as a function of subsequent time at 200 °C.

Measurement versus anneal temperature for Ti/Al/ZrB₂/Ti/Au on n-GaN. A) Specific contact resistivity. B) Sheet resistance.

Contact resistance of the Ti/Al/W₂B/Ti/Au on n-GaN, initially annealed at 800 °C, as a function of subsequent time at 200 °C.

Measurement versus anneal temperature for Ti/Al/ZrB₂/Ti/Au on n-GaN. A) Specific contact resistivity. B) Sheet resistance.

Measurement as a function of annealing time at 700 °C for Ti/Al/ZrB₂/Ti/Au on n-GaN. A) Specific contact resistivity. B) Sheet resistance.

Specific contact resistance versus measurement temperature for Ti/Al/ZrB₂/Ti/Au on n-GaN annealed at 800 °C.

Secondary electron images of the Ti/Al/ZrB₂/Ti/Au on n-GaN. A) As-deposited. B) Annealed at 500 °C. C) Annealed at 700 °C. D) Annealed at 1000 °C.

AES surface scans of the Ti/Al/ZrB₂/Ti/Au on n-GaN. A) As-deposited. B) Annealed at 500 °C. C) Annealed at 700 °C. D) Annealed at 1000 °C.

AES depth profiles of the Ti/Al/W₂B/Ti/Au on n-GaN. A) As-deposited. B) Annealed at 500 °C. C) Annealed at 800 °C. D) Annealed at 1000 °C.

Secondary electron images of the Ti/Al/ZrB₂/Ti/Au on n-GaN. A) As-deposited. B) Annealed at 500 °C. C) Annealed at 700 °C. D) Annealed at 1000 °C.

AES surface scans of the Ti/Al/ZrB₂/Ti/Au on n-GaN. A) As-deposited. B) Annealed at 500 °C. C) Annealed at 700 °C. D) Annealed at 1000 °C.

AES depth profiles of the Ti/Al/ZrB₂/Ti/Au on n-GaN. A) As-deposited. B) Annealed at 500 °C. C) Annealed at 700 °C. D) Annealed at 1000 °C.

AES depth profiles of CrB₂-based contacts. A) As-deposited. B) Annealed at 700 °C. C) Annealed at 800 °C. D) Annealed at 1000 °C.

AES depth profiles of TiB₂-based contacts. A) As-deposited. B) Annealed at 600 °C. C) Annealed at 800 °C. D) Annealed at 1000 °C.

AES depth profiles of W₂B₅-based contacts. A) As-deposited. B) Annealed at 500 °C. C) Annealed at 700 °C. D) Annealed at 1000 °C.

Specific contact resistance of the boride-based contacts annealed at 800 °C and the conventional Ti/Al/Ni/Au contacts as a function of aging time at 350 °C.
5-1 SEM micrographs of W$_2$B based schottky contacts. A) As-deposited. B) Annealed at 700 °C. The inner circle is the W$_2$B/Ti/Au while the outer ring is the Ohmic contact. ...95

5-2 Barrier height and reverse breakdown voltage as a function of measurement temperature for as-deposited W$_2$B/Ti/Au contacts on n-GaN. ...........................................96

5-3 Barrier height and reverse breakdown voltage as a function of annealing temperature for W$_2$B/Ti/Au contacts on n-GaN. .................................................................97

5-4 AES depth profiles of W$_2$B/Ti/Au on GaN. A) Unannealed. B) After annealing at 700 °C. ..........................................................................................................................98

5-5 I-V characteristics from ZrB$_2$/GaN diodes as a function of post-deposition annealing temperature. .............................................................................................................99

5-6 Barrier height and reverse breakdown voltage as a function of annealing temperature for ZrB$_2$/Ti/Au contacts on n-GaN. ..............................................................100

5-7 SEM micrographs of ZrB$_2$ based schottky contacts. A) As-deposited. B) Annealed at 350 °C. C) Annealed at 700 °C ........................................................................................................101

5-8 AES surface scans of ZrB$_2$/Ti/Au on GaN. A) As-deposited. B) Annealed at 350 °C. C) Annealed at 700 °C. ........................................................................................................102

5-9 AES depth profiles of ZrB$_2$/Ti/Au on GaN. A) As-deposited. B) Annealed at 350 °C. C) Annealed at 700 °C. ........................................................................................................103

5-10 Powder XRD spectrum from ZrB$_2$ on GaN. A) Unannealed. B) After annealing at 800 °C. .................................................................................................................................104

5-11 Glancing angle XRD spectra from ZrB$_2$ on GaN. A) Unannealed. B) After annealing at 800 °C. .................................................................................................................................105

6-1 I-V characteristics at 25° C of TiB$_2$/Ti/Au on GaN as a function of post-deposition annealing temperature........................................................................................................117

6-2 Barrier height and reverse breakdown voltage as a function of annealing temperature for TiB$_2$/Ti/Au contacts on n-GaN..................................................................................118

6-3 AES surface scans of TiB$_2$/Ti/Au on GaN. A) As-deposited. B) Annealed at 350 °C. C) Annealed at 700 °C. ........................................................................................................119

6-4 AES depth profiles of TiB$_2$/Ti/Au on GaN. A) As-deposited. B) Annealed at 350 °C. C) Annealed at 700 °C. ........................................................................................................120

6-5 SEM micrographs of TiB$_2$ based schottky contacts A) As-deposited. B) Annealed at 350 °C. C) Annealed at 700 °C ........................................................................................................121
6-6 Barrier height and reverse breakdown voltage as a function of measurement temperature for as-deposited TiB$_2$/Ti/Au contacts on n-GaN.

6-7 SEM micrographs of W$_2$B$_5$ based schottky contacts. A) As-deposited. B) Annealed at 350 °C.

6-8 AES depth profiles of W$_2$B$_5$/Ti/Au on GaN. A) As-deposited. B) Annealed at 350 °C. C) Annealed at 700 °C.

6-9 I-V characteristics of W$_2$B$_5$/Ti/Au on GaN as a function of post-deposition annealing temperature.

6-10 Barrier height and reverse breakdown voltage as a function of annealing temperature for W$_2$B$_5$/Ti/Au contacts on n-GaN.

6-11 I-V characteristics of as-deposited W$_2$B$_5$/Ti/Au on GaN as a function of measurement temperature.

6-12 Barrier height and reverse breakdown voltage as a function of measurement temperature for as-deposited W$_2$B$_5$/Ti/Au contacts on n-GaN.

6-13 I-V characteristics of CrB$_2$/Ti/Au on GaN as a function of post-deposition annealing temperature.

6-14 Barrier height and reverse breakdown voltage as a function of annealing temperature for CrB$_2$/Ti/Au contacts on n-GaN.

6-15 AES depth profiles of CrB$_2$/Ti/Au on GaN. A) As-deposited. B) Annealed at 350 °C. C) Annealed at 700 °C.

6-16 AES surface scans of CrB$_2$/Ti/Au on GaN. A) As-deposited. B) Annealed at 350 °C. C) Annealed at 700 °C.

6-17 SEM micrographs of CrB$_2$ based schottky contacts. A) As-deposited. B) Annealed at 350 °C. C) Annealed at 700 °C.

6-18 Barrier height and reverse breakdown voltage as a function of measurement temperature for as-deposited CrB$_2$/Ti/Au contacts on n-GaN.

7-1 HEMT layout used in these experiments.

7-2 Study of Raman spectra from Ti/Al/TiB$_2$/Ti/Au contacts on HEMT wafer. A) Optical micrograph. B) Raman spectra.

7-3 I$_{DS}$-V$_{DS}$ characteristics from HEMT with conventional Pt/Au gate contacts and Ti/Al/Pt/Au source/drain contacts before and after aging at 350 °C for 25 days.

7-4 I$_{DS}$-V$_{DS}$ characteristics from HEMT with Ti/Al/TiB$_2$/Ti/Au source/drain contacts. A) Pt/Au gate contacts before and after aging at 350 °C for 25 days.
7-5 $I_{DS}-V_{DS}$ characteristics from HEMT with Ti/Al/TiB$_2$/Ti/Au source/drain contacts. A) Ni/Au gate contacts before and after aging at 350 °C for 25 days. ........................................144

7-6 Optical microscopy images of HEMTs. A) Conventional contacts before aging. B) Boride-based source/drain contacts before aging. ............................................................145

7-7 Percent change in saturated drain/source current from HEMTs with different combinations of contact metal schemes as a function of aging time at 350 °C..............146

7-8 RF performance of 1.5 \times 200 \mu m^2 gate length HEMTs. A) HEMT with conventional metal contacts prior to aging..........................................................147

8-1 I-V characteristics from Ir/Au Schottky contacts on n-GaN. ...........................................154

8-2 Schottky barrier height for Ir/Au contacts on n-GaN as a function of annealing temperature. ...............................................................................................................155

8-3 AES depth profiles. A) Ir/Au after annealing at 350 °C. B) Ir/Au after annealing at 700 °C. C) Ni/Au contacts after annealing at 350 °C......................................................156

8-4 Specific contact resistance of Ti/Al//Ni/Au and Ti/Al/Ir/Au Ohmic contacts on n-GaN as a function of anneal temperature.........................................................157

8-5 SEM images Ir and Ni based ohmic. A) Ti/Al/Ir/Au after annealing at 500 °C. B) Ti/Al/Ni/Au after annealing at 500 °C.................................................................158

8-6 AES depth profiles of Ir and Ni based ohmic. A) Ti/Al/Ir/Au after annealing at 500 °C. B) Ti/Al/Ni/Au after annealing at 500 °C ......................................................159

8-7 Specific contact resistance of Ti/Al//Ni/Au and Ti/Al/Ir/Au Ohmic contacts on n-GaN as a function of measurement temperature...........................................160

8-8 Specific contact resistance of the Ti/Al//Ni/Au and Ti/Al/Ir/Au contacts annealed at 900°C as a function of aging time at 350 °C.............................................161
In this work the effort was made to develop and investigate high temperature stable Ohmic and Schottky contacts for n type GaN. Various borides and refractory materials were incorporated in metallization scheme to best attain the desired effect of minimal degradation of contacts when placed at high temperatures.

This work focuses on achieving a contact scheme using different borides which include two Tungsten Borides (namely W$_2$B, W$_2$B$_5$), Titanium Boride (TiB$_2$), Chromium Boride (CrB$_2$) and Zirconium Boride (ZrB$_2$). Further a high temperature metal namely Iridium (Ir) was evaluated as a potential contact to n-GaN, as part of continuing improved device technology development. The main goal of this project was to investigate the most promising boride-based contact metallurgies on GaN, and finally to fabricate a High Electron Mobility Transistor (HEMT) and compare its reliability to a HEMT using present technology contact.

Ohmic contacts were fabricated on n GaN using borides in the metallization scheme of Ti/Al/boride/Ti/Au. The characterization of the contacts was done using current-voltage measurements, scanning electron microscopy (SEM) and Auger Electron Spectroscopy (AES) measurements. The contacts formed gave specific contact resistance of the order of $10^{-5}$ to $10^{-6}$ Ohm-cm$^2$. A minimum contact resistance of $1.5 \times 10^{-6}$ $\Omega$.cm$^2$ was achieved for the TiB$_2$ based
scheme at an annealing temperature of 850-900 °C, which was comparable to a regular ohmic contact of Ti/Al/Ni/Au on n GaN. When some of borides contacts were placed on a hot plate or in hot oven for temperature ranging from 200 °C to 350 °C, the regular metallization contacts degraded before than borides ones. Even with a certain amount of intermixing of the metallization scheme the boride contacts showed minimal roughening and smoother morphology, which, in terms of edge acuity, is crucial for very small gate devices.

Schottky contacts were also fabricated and characterized using all the five boride compounds. The barrier height obtained on n GaN was ~0-5-0.6 eV which was low compared to those obtained by Pt or Ni. This barrier height is too low for use as a gate contact and they can only have limited use, perhaps, in gas sensors where large leakage current can be tolerated in exchange for better thermal reliability.

AlGaN/GaN High Electron Mobility Transistors (HEMTs) were fabricated with Ti/Al/TiB$_2$/Ti/Au source/drain ohmic contacts and a variety of gate metal schemes (Pt/Au, Ni/Au, Pt/TiB$_2$/Au or Ni/TiB$_2$/Au) and were subjected to long-term annealing at 350 °C. By comparison with companion devices with conventional Ti/Al/Pt/Au ohmic contacts and Pt/Au gate contacts, the HEMTs with boride-based ohmic metal and either Pt/Au, Ni/Au or Ni/TiB$_2$/Au gate metal showed superior stability of both source-drain current and transconductance after 25 days aging at 350 °C.

The need for sputter deposition of the borides causes’ problem in achieving significantly lower specific contact resistance than with conventional schemes deposited using e-beam evaporation. The borides also seem to be, in general, good getters for oxygen leading to sheet resistivity issues.
Ir/Au schottky contacts and Ti/Al/Ir/Au ohmic contacts on n-type GaN were investigated as a function of annealing temperature and compared to their more common Ni-based counterparts. The Ir/Au ohmic contacts on n-type GaN with n~ 10^{17} cm^{-3} exhibited barrier heights of 0.55 eV after annealing at 700 °C and displayed less intermixing of the contact metals compared to Ni/Au. A minimum specific contact resistance of 1.6 \times 10^{-6} \, \Omega \cdot cm^2 was obtained for the ohmic contacts on n-type GaN with n~10^{18} \, cm^{-3} after annealing at 900 °C. The measurement temperature dependence of contact resistance was similar for both Ti/Al/Ir/Au and Ti/Al/Ni/Au, suggesting the same transport mechanism was present in both types of contacts. The Ir-based ohmic contacts displayed superior thermal aging characteristics at 350 °C. Auger Electron Spectroscopy showed that Ir is a superior diffusion barrier at these moderate temperatures than Ni.
CHAPTER 1
INTRODUCTION

The microelectronics industry has grown rapidly in the past four decades and now is the basis for our Information Age. The first semiconductor transistor was invented by the scientists of Bell Labs in 1947. Subsequently, the concept of an Integrated Circuit (IC) was developed, requiring a high yield of working devices that comprise the circuit. To have a 50% probability of functionality for a 20 transistor circuit, the probability of device functionality must be \((0.5)^{1/20} = 0.966\) or 96.6%. This was considered wildly optimistic at the time, yet today integrated circuits are built with billions of transistor. This is possible because each component or a device is many times reliable compared to a component in any other industry. Even though the very first semiconductor transistor was made from germanium (Ge), silicon (Si) became the semiconductor of choice as a result of the low melting point of Ge that limits high temperature processes and the lack of a natural occurring germanium oxide to prevent the surface from electrical leakage.

Due to the maturity of its fabrication technology, silicon continues to dominate the present commercial market in discrete devices and integrated circuits for computing, power switching, data storage and communication. For high-speed and optoelectronic devices such as high-speed integrated circuits and laser diodes, gallium arsenide (GaAs) is the material of choice. It exhibits superior electron transport properties and special optical properties. GaAs has higher carrier mobility and higher effective carrier velocity than Si, which translate to faster devices. GaAs is a direct bandgap semiconductor, whereas Si is indirect, hence making GaAs better suited for optoelectronic devices. However, physical properties required for high power, high temperature electronics and UV/blue light emitter applications are beyond the limits of Si and GaAs. It is essential to investigate alternative materials and their growth and processing techniques in order to achieve these devices. So now the focus has shifted to semiconductors having wide
bandgaps. They exhibit inherent properties such as larger bandgap, higher electron mobility and higher breakdown field strength making them suitable for high power, high temperature electronic devices and short wavelength optoelectronics.

Wide bandgap semiconductors offer the best technical promise for high power and high temperature transistors. Until recently, the most promising of these materials was silicon carbide (SiC). However, SiC has several technical shortfalls that have opened competition to the III-nitride materials. Thermal oxides in SiC power metal oxide semiconductor field effect transistors (MOSFETs) actually limit the temperature range of application since the gate contact degrades and becomes electrically leaky at high temperatures. The low electron mobility of only 400 cm$^2$/V.s yields lower PAE (<30%) for many transistors in the frequency range of 1 to 5 GHz. For silicon and SiC, amplifier efficiency decays rapidly with increase in frequency so that it drops below 25% for many devices operating above 2 GHz. GaN-base devices offer wider bandgap, greater chemical inertness and higher temperature stable operation than SiC.

Single transistor output power is the most important cost limiting issue for commercialization of solid-state power devices. Other economic factors relating to performance are power-added efficiency (PAE) required for lightweight portable systems, amplifier linearity necessary to transmit digital signals without distortion or out-of-band modulation products, and amplifier noise figure and phase noise. Output power achievable by microwave devices is directly proportional to the breakdown voltage and sustainable current limits. For bipolar devices under Class A operation, the maximum output power density is then

$$P_{\text{max}} = \frac{I_{\text{sat}} (V_{\text{cb}} - V_{\text{knee}})}{8}$$

where $I_{\text{sat}}$ is the saturation current at the quiescent point, $V_{\text{cb}}$ is the collector breakdown voltage and $V_{\text{knee}}$ is the saturation voltage at maximum current. Apart from high breakdown voltage, high thermal stability, GaN-based semiconductors also benefit from a very high sustainable electron
saturation velocity of 2.7x10^7 cm/s. This unique property, which has been shown to significantly benefit GaN FETs, is the result of large energetic displacement between valleys in the conduction band profiles 120.

One of the most significant problems limiting single transistor high power devices is the heat dissipation required. Mature silicon RF power transistors are currently limited to about 125 °C junction temperature (85-100 °C ambient) with operation of little more than 1 W at 10 GHz. Due to leaky oxides, SiC does not increase this range enough to result in significant advantage. GaAs technology has improved on this performance to yield 50 W at 10 GHz with state-of-the-art power FET technology. However, both silicon and GaAs devices suffer greater high temperature de-rating than is expected from the wide bandgap GaN devices. The GaN devices not only can operate at 400 °C or higher but also should exhibit optimal performance somewhere near 250 °C due to improved ionization of the carriers in the material 119.

While further improvements in the III-V nitride materials quality can be expected to enhance device operation, further device advances will also require improved processing technology. Owing to their wide bandgap nature and chemical stability, GaN and related materials present a host of device processing challenges, including difficulty in achieving reliable low-resistance ohmic contacts, thermally stable contacts for both n and p GaN, high temperatures needed for implant activation, lack of efficient wet etch process, generally low dry etch rates and low selectivity over etching masks, and dry etch damage. High thermal budget and dry etch damage indirectly adds to the problem of having good reliable ohmic and schottky contacts. These problems constitute a major obstacle to successful demonstration and commercialization of some GaN-based devices, such as bipolar transistors and power switches, whose performance are much more affected by the immature fabrication techniques. To fully
exploit these device applications, a number of critical advances are necessary\(^4\,119\). One of the
critical area is high temperature thermal processing, ohmic and schottky contacts which are
thermally stable and can at least sustain harsh condition which the device itself is capable of based
on its intrinsic properties.

The motivation of this work is to develop novel ohmic and schottky contacts to GaN and
AlGaN/GaN high electron mobility devices for use in high temperature application. So the
objective of this work is to have high temperature stable ohmic and schottky contact to n-GaN
which should circumvent or delay the problem of intermixing of metal layers and surface
roughening leading to a better and reliable contact scheme. In this project, we explore a novel
metallization scheme involving borides because of the refractory nature of the borides and thus
thermal stability and very little possibility of it having solid state reactions with other metals
normally used in contact scheme. Apart from boride, a high temperature metal, namely Ir, was
also explored as part of continuing search for better contacts. The objective is to have an
optimized new contact for high temperature operation of AlGaN/GaN HEMTs.

The properties of GaN and background of semiconductor processing and characterizations
especially in terms of ohmic and schottky contacts are reviewed in Chapter 2. Ohmic and
schottky contacts are necessary to impart specific electrical interactions and characteristics in
achieving operating devices. The studies of ohmic and schottky contact metallization are covered
in Chapters 3 through 7. The Tungsten Boride (W\(_2\)B) and Zirconium Boride (ZrB\(_2\)) metal
scheme was considered first. Next a comparative study of Tungsten Boride (W\(_2\)B\(_3\)), Titanium
Boride (TiB\(_2\)) and Chromium Boride (CrB\(_2\)) is done and is presented in chapter 4. Chapter 5
discusses the rectifying nature of Tungsten Boride (W\(_2\)B) and Zirconium Boride. Chapter 6
shows result of Titanium Boride, Tungsten Boride (W\(_2\)B\(_3\)) and Chromium Boride in usage as a
rectifying contact to n GaN. The demonstration of a High Electron Mobility Transistor using new best boride based metallurgy is given in Chapter 7. Chapter 8 deals with high melting temperature metal Iridium (Ir) as being explored as ohmic and schottky contact to n GaN. The conclusion and summary of the study of new boride based and Ir contacts to n GaN and HEMTs is given in chapter 9.
Overview

GaN is a wide bandgap semiconductor which has numerous properties which makes it well suited for high temperature applications. Its electrical properties are compared to Si, SiC and other materials in Table 2-1. It has a direct bandgap energy of 3.45 eV (\(\lambda=359.37\) nm) which is transparent to visible light and operates in ultra violet to blue wavelengths. Hall measurements at room temperatures show the Hall mobility of electron of 1000~1300 cm\(^2\)/V-s. It has saturation velocity little higher than GaAs. GaN like ZnO seems to be extremely stable at harsh environment of gamma radiations. It has little change in IV characteristic even after being irradiated by high energy proton radiation. This makes GaN very good candidate for outer space and nuclear application. Sapphire or SiC substrates are generally used for growing GaN.

GaN also has different heterostructures available with Al, In etc. (Al, Ga, In) N forms a continuous and direct band gap alloy from 1.92 eV (InN) to 6.2 eV (AlN) with potential for emission and detection in spectral range between visible and the ultraviolet wavelengths.

Crystal Structure and Basic Properties

GaN is a direct bandgap semiconductor having stable form as hexagonal (wurtzite) crystal structure, with lattice parameters \(a = 3.189\) Å and \(c = 5.178\) Å. The Ga (group III) atoms are tetrahedrally coordinated with four N (group V) atoms. Alternating Ga and N layers form the crystal structure (Figure 2-1).

GaN-based semiconductors have attracted tremendous interest for their applications to blue laser and LEDs, high temperature, high power electronics, high density optical data storage, and
electronics for the aerospace and automobile industries, telecommunication devices, and wide band gap semiconductors in power amplifiers extends the radiation hardness of the circuit \(^7\).

Many of these compounds are shown graphically in Figure 2-2 \(^{10}\) in terms of their crystallographic lattice constant versus the energy band gap. Especially, wide band gap electronic devices have excellent electrical and physical characteristics. Table 2-2 shows the physical parameters in different semiconductor materials. The high power, high frequency operation most promising materials are GaN and SiC and the band gap energy is 3.4 eV and 3.2 eV respectively. For example, to get \(10^{15}/\text{cm}^3\) intrinsic carrier concentration \((n_i)\), we need 300 °C for the Si materials, 500 °C for the GaAs, however much higher temperatures are needed to get the same intrinsic carrier concentration in the GaN, namely about 1000 °C. For these reasons, the GaN is much better for use in high temperature conditions, and devices made out of it will operate more reliably at elevated temperature.

The early unintentionally doped GaN was n-type, which at that time was believed due to nitrogen vacancies. The high n-type background carrier concentration on the order of \(10^{18} \text{ cm}^{-3}\) proved difficult to minimize and the absence of a shallow acceptor dimmed the prospects of a production-scale GaN-based device effort.

Table 2-3 shows the ionization energy of impurities for GaN. Si is the most general n-type dopant of for GaN since it effectively incorporates on the Ga site and forms a single shallow donor level. Si is fully ionized at room temperature with the ionization level of \(~30\text{ meV}\). The early unintentionally doped GaN was n-type, which at that time was believed due to nitrogen vacancies. The high n-type background carrier concentration on the order of \(10^{18} \text{ cm}^{-3}\) proved difficult to minimize and the absence of a shallow acceptor dimmed the prospects of a production-scale GaN-based device effort.

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Basic physical properties of GaN are listed into Table 2-4 \(^5,6\). These values are result of various works and some values have uncertainty because of the fact that different materials are used for experiments and there remains certain inhomogeneity. There is also a metastable form of GaN as Zinc Blend structure. The variations in different properties like calculated mobility,
thermal conductivity are possible because of crystal defects such as dislocations. Defects in the materials are very critical factor in the effect. Lots of dislocations are caused by lattice mismatch, which is 13% on the sapphire, 3% on the SiC substrate.

The successes of all GaN related devices depend largely on having excellent ohmic and schottky contacts to these devices. A figure of a GaN HEMT is shown in Figure 2-3. There are two types of contacts to a semiconductor. One contact is ohmic and other is schottky

**Metal Contact**

At present improvement in contact has become a critical factor for better technology along with advancing the properties of the semiconductors itself. In recent years, GaN itself have been proven to be excellent choice for high temperature, high frequency applications. The successes of all GaN related devices for high temperature application will depend largely on having excellent contacts to these devices. There are two types of contacts to a semiconductor. Contact to semiconductor basically consists of region of semiconductor surface just below first metal layer, metal semiconductor interface and few layers of metallization above it. Invariably the as deposited contact does not give the desired properties (either low resistance or high schottky barrier). So the contacts are annealed which results in formation of different complex intermetallic compounds by way of solid state reaction among metal layers and semiconductor surface. Thus a contact simply is referred to as the region of metal semiconductor interface that leads to desirable electrical characteristic. The current transport in metal-semiconductor contact occurs by majority carriers. There are two different types of contacts namely ohmic and schottky. In ohmic contact the current-voltage relation follows Ohms law that is it should be linear. The contact resistance should be very low so that there is negligible voltage drop across it and hence negligible power drop. This is very important for devices and more so in power application where minimum loss and maximum efficiency is required. Another critical requirement for high
temperature application is the need to have contact which does not degrade or rather have a high resistance to degradation. Smooth surface morphology, sharp edge acuity and reliability and reproducibility are other features that are desired in an ideal contact.

Another type of contact is schottky contact, or rectifying contact in which large current can flow in one direction at small voltage and almost no current in reverse direction. High barrier height is essential for producing rectifying effects.

Whether a metal-semiconductor interface forms an ohmic or schottky contact depends upon the metal work function, $\phi_m$, and semiconductor work function, $\phi_s$. Work function is the amount of energy required to excite an electron from Fermi energy level to the vacuum level. Theoretically, on n-type semiconductor, ohmic contact is formed when $\phi_m < \phi_s$, and schottky contact is formed when $\phi_m > \phi_s$. Conversely, in p-type material, $\phi_m > \phi_s$ and $\phi_m < \phi_s$ produces ohmic and rectifying contact, respectively. Selected values of work function for commonly used metals are shown in Table 2-5. The semiconductor work function is sum of the electron affinity and energy difference between Fermi energy and the bottom of the conduction band i.e. $\phi_s = \chi_s + \xi$, where $\chi_s$ is the electron affinity and $\xi$ is the energy difference between the Fermi energy and the conduction band. The electron affinity for GaN is 4.1 eV. The work function of Tungsten (W), Cr, Ti, Zr is 4.55, 4.5, 4.33 and 4.05 eV respectively.

**Schottky contact**

When an intimate contact is formed between metal and a semiconductor, the Fermi levels in the two materials must be coincident at thermal equilibrium. This can be achieved through a charge flow from semiconductor to metal. Thus a barrier forms at the interface and an equal and opposite space charge is distributed over the barrier region near the semiconductor surface. With an n-type semiconductor in the absence of surface state, the barrier height $q\phi_{bn}$ is given by
\[ q \phi_m = q(\phi_m - \chi) \]  

(2-1)

where \( q \phi_m \) is the metal work function, \( q \chi \) is the electron affinity of the semiconductor. For an ideal contact between metal and a p-type semiconductor, the barrier height \( q \phi_{bp} \) is given by

\[ q \phi_{bp} = E_g - q(\phi_m - \chi) \]  

(2-2)

When surface states are present on the semiconductor surface, and the density is sufficiently large to accommodate any additional surface charges without appreciably altering the occupation level \( E_F \), the space charge in the semiconductor will remain unaffected. As a result, the barrier height is determined by the property of the semiconductor surface, and is independent of the metal work function. In practice, some surface states are always present at the semiconductor surface, and continuously distributed in energy within the energy gap. The schottky barrier heights of metal-semiconductor systems with intimate contact are, in general, determined by both the metal work function and the surface states.

In a simple model for all semiconductors, the schottky barrier height \( q \phi_b \) can be expressed as

\[ q \phi_b = q(S\chi_m + \phi_0) \]  

(2-3)

where \( \chi_m \) is metal electronegativity, \( \phi_0 \) represents the contribution of surface states of semiconductors, and interface index \( S = \frac{d \phi_b}{d \chi_m} \), is found to be a function of the electronegativity difference \( \Delta \chi \) between cation and anion of compound semiconductor (Figure 2-4 A). Note a sharp transition around \( \Delta \chi = 1 \). For ionic semiconductors, \( \Delta \chi > 1 \), the index \( S \) approaches 1, and \( \phi_b \) is strongly dependent of the metal electronegativity (or work function). On the other hand, for covalent semiconductors with \( \Delta \chi < 1 \), \( S \) is small, \( \phi_b \) is affected by high density surface states from dangling bonds and only weakly depends on metal work function. GaN has an electronegativity difference of 1.4 (Ga: 1.6, N: 3.0), which would predict the schottky barrier heights depend on
metal work function, and are given by Equation 2-1 and Equation 2-2 for metal on n-type and p-type material respectively. A summary of reported schottky barrier heights of a variety of elemental metals on n-GaN is shown in Figure 2-4 B \(^{12}\). It is clear that the barrier height indeed varies with the metal work function within experimental scattering.

The current transport in metal-semiconductor contacts is mainly due to majority carrier, in contrast to p-n junctions. Two major processes under forward bias are (1) transport of electrons from the semiconductor over the potential barrier into the metal; (2) quantum-mechanical tunneling of electrons through the barrier. In addition, we may have recombination current in the space-charge region and leakage current at the contact periphery. The transport of electrons over the potential barrier is often the dominant process for schottky diodes on moderately doped semiconductors. It can be adequately described by thermionic emission theory for high mobility semiconductor (for low mobility materials, the diffusion theory is also applicable), and the electric current density over the barrier has the following expression:

\[
J = J_s \left[ \exp\left(\frac{qV}{kT}\right) - 1 \right] = A^{**}T^2 \exp\left(-\frac{q\phi_b}{kT}\right)[\exp\left(\frac{qV}{kT}\right) - 1]
\]

(2-4)

where \(J_s\) is the saturation current density, \(A^{**}\) is the effective Richardson constant. In practical device, the barrier height dependent on bias voltage and the current-voltage characteristics is more accurately described by:

\[
J = A^{**}T^2 \exp\left(-\frac{q\phi_b}{kT}\right)[\exp\left(\frac{qV}{nkT}\right) - 1]
\]

(2-5)

Factor \(n\) is called the ideality factor. The barrier height and ideality factor can be obtained from the forward J-V characteristics (for \(V>3kT/q\)):

\[
n = \frac{q}{kT} \frac{\partial V}{\partial \ln J}
\]

(2-6)

\[
\phi_b = \frac{kT}{q} \ln\left(\frac{A^{**}T^2}{J_s}\right)
\]

(2-7)
For a heavily doped semiconductor or for operation at low temperatures, the tunneling current may become the dominant transport process. The tunneling current has an expression:

\[ J_t \sim \exp \left( \frac{2\phi_h}{h} \sqrt{\frac{\varepsilon_s m^*}{N_D}} \right) \]

where \( \varepsilon_s \) is permittivity of semiconductor, \( m^* \) is effective mass of carrier, \( N_D \) is carrier concentration. It indicates the current will increase exponentially with \( N_D^{0.5} \).

Earlier work on schottky contacts to n-GaN have been done based on metals layers consisting mainly of Ni or Pt with Au above it \[37-40, 65, 66\]. W/Ti/Au and WSi\(_x\)/Ti/Au schemes have also been used as schottky resulting in thermally stable schottky with barrier height of \(~0.80\) eV which reduced to \(~0.4\) eV for subsequent annealing at 400 °C \[68\]. The barrier height seems to follow the difference in work function value with in experimental scattering.

Schottky diode can be made by depositing an inner circular schottky metal scheme with an outer concentric ring as the ohmic metal scheme. The outer ohmic metal is first deposited and annealed to get the desired ohmic characteristic and inner circle is realigned. A diagram of a schottky structure is shown is Figure 2-5.

**Ohmic contact**

It is imperative that a semiconductor device be connected to the outside world with no adverse change to its current-voltage characteristics. This can be accomplished through ideal ohmic contacts to the semiconductor. An ohmic contact is defined as a metal/semiconductor contact that has negligible contact resistance relative to the bulk or spreading resistance of the semiconductor. A satisfactory ohmic contact can supply the required current with a voltage drop that is sufficiently small compared with the drop across the active region of the devices. One important figure of merit for ohmic contact is specific contact resistance \( r_c \), which is defined as:
\[ r_c = \left( \frac{\partial I}{\partial V} \right)_{V=0}^{-1} \tag{2-9} \]

For contact with lower doping concentration, at relatively high temperature, conduction across the M/S interface is dominated by thermionic-emission over the potential barrier, as given in Equation 2-4. Therefore,

\[ r_c = \frac{k}{qA^*T} \exp \left( \frac{q\phi_b}{kT} \right) \tag{2-10} \]

It is obvious that low \( \phi_b \) should be used for small \( r_c \). Ideally a metal with a lower work function than an n-type semiconductor or higher work function than a p-type semiconductor should be used for ohmic contact to this semiconductor. Unfortunately, very few practical material systems satisfy this condition, and metals usually form schottky barriers at semiconductor interface. A practical way to obtain a low resistance ohmic contact is to create a highly doped region near the surface by ion implantation, or increase the doping by alloying the contacts. In this case, the depletion layer cause by the schottky barrier becomes very thin, and current transport through the barrier is enhanced by tunneling. The contact resistance can be obtained from Equation 2-8,

\[ r_c \sim \exp \left( \frac{2\sqrt{\varepsilon m^* \phi_b}}{h \sqrt{N_D}} \right) \tag{2-11} \]

Note that \( r_c \) depends strongly on \( N_D \). Under intermediate conditions, thermionic field emission is important, where there is enough kinetic energy for the carrier to be excited to an energy level at which the potential barrier is thin enough for tunneling to occur. Typical ohmic conduction is usually related to a large tunneling component.

It is difficult to make ohmic contacts on wide-bandgap semiconductors, such as GaN (\( \varepsilon_g = 3.4 \text{ eV}, \chi = 4.1 \text{ eV} \)) and SiC. Generally the doping concentration is relatively low due to the high ionization level of typical dopants.
A wide variety of metallization schemes been tried for ohmic contact for n-GaN. Some of the earliest report of ohmic contact to n-GaN had Al as the ohmic contact metal with specific contact resistivity of ~ $10^{-7}$ $\Omega \cdot \text{cm}^2$. Specific contact resistivity of $8 \times 10^{-6}$ $\Omega \cdot \text{cm}^2$ using Ti/Al was achieved after 900 $^\circ$C anneal for 30 sec in N$_2$ ambient. Ti/Al contact with Si implantation resulted in specific contact resistivity of $3.6 \times 10^{-8}$ $\Omega \cdot \text{cm}^2$ and with RIE pre treatment resulted in $8.9 \times 10^{-8}$ $\Omega \cdot \text{cm}^2$. A Specific contact resistivity of $~ 8 \times 10^{-5}$ $\Omega \cdot \text{cm}^2$ was reported for W on n-GaN. Specific contact resistivity of $5.6 \times 10^{-6}$ $\Omega \cdot \text{cm}^2$ for Al/Ti contact on AlGaN/GaN hetrostructure with Si implantation, $5.3 \times 10^{-7}$ $\Omega \cdot \text{cm}^2$ for Ta/Ti/Al contacts, $1.2 \times 10^{-5}$ $\Omega \cdot \text{cm}^2$ for Ti/Al/Ni/Au contacts have been also reported. The most common contact scheme used is Ti/Al bilayer with Ni/Au, Ti/Au/ or Pt/Au over layer where overlayer is mainly for preventing out diffusion, smooth morphology and Au is used for reducing sheet resistance of the layers and to prevent oxidation during high temperature anneal. High temperature metals have also been used for ohmic schemes to have better long term stability. Ti/Al/Mo/Au, Ti/Al/Ir/Au, W and WSi$_x$ gave specific contact resistivity of $\sim 10^{-5}$ $\Omega \cdot \text{cm}^2$.

Contact resistance is measured and Specific contact resistance is determined by Transfer length model (TLM), also known as transmission line model. Linear TLM patterns consist of square or rectangular contact pads separated by different spacing. There is also a Circular TLM patterns which has concentric circular metal patterns where either the inner radii or the outer radii change to vary gap distance. Schematics of linear TLM and measurement plot is shown in Figures 2-6.

Current and voltage information obtained from electrical measurements are curve fitted with the corresponding equations to determine the specific contact resistance. For linear TLM, the total resistance, $R_s$, and specific contact resistance, $\rho_c$, are given by
where $R_C$ is the contact resistance, $R_s$ is the sheet resistance, $L$ is the distance between two pads, $W$ is the width of the pad. For Circular TLM, the specific contact resistance, $\rho_c$, can obtained form the circular TLM measurements with the relationships \textsuperscript{108, 120}

$$R_T = \frac{R_S}{2\pi} \left[ \ln \left( \frac{R_1}{R_O} \right) + \frac{L_T}{R_O} K_0 \left( \frac{R_1}{L_T} \right) + \frac{L_T}{R_1} K_1 \left( \frac{R_1}{L_T} \right) \right]$$

$$\rho_c = \frac{R_T^2 W^2}{R_S}$$

where $R_T$ is the total resistance, $R_S$ is the sheet resistance, $R_1$ is the outer radius of the annular gap, $R_O$ is the inner radius of the annular gap, $I_O$, $I_1$, $K_O$, and $K_1$ are the modified Bessel functions, $L_T$ is the transfer length, and $\rho_c$ is the specific contact resistance.

**Thermal stability**

Thermal processing such as activation of ion implants and alloying of metal contacts can be detrimental to device operation due to changes in the material, interaction, or reactions, as also observed in GaAs \textsuperscript{109-111, 120}. It is very important to have contacts that are able to resist the high temperature long enough to be commercially possible for high temperature applications. In regards to this a good understanding of the degradation of the material is helpful in identifying high temperature process limits. Reliable and stable operation of devices largely depends upon the thermal stability of the contacts. At high temperature a lot of intermetallics compounds may form in contacts as a result of interdiffusion of different metal. This can result in rough surface morphology, change in stoichiometry, change in composition resulting in change in electrical and optical properties.
Common Processing Techniques

**Dry Plasma Etching**

Etching refers to the crucial IC fabrication process of transferring pattern by removing specified areas. Wet chemical etching was widely used in manufacturing until the 1960s. Even though this technique is inexpensive, the feature size is limited to about 3 microns. The isotropic etching results in sloped sidewalls and undercutting of the mask material. As feature dimension decreases to microns and submicrons and device density per chip increases, anisotropic etching is necessary. Dry etching techniques using gases as primary etch medium were developed to meet this need. In addition to anisotropic pattern transfer, dry etching provides better uniformity across the wafer, higher reproducibility, smoother surface morphology, and better control capability than wet chemical etching. Three general types of dry etching include plasma etching, ion beam milling, and reactive ion etch (RIE)\textsuperscript{112, 113, 120}. Inductively coupled plasma (ICP) etching was used in this study and will be discussed in detail.

ICP etching is a dry etching technique where high-density plasmas are formed in a dielectric vessel encircled by inductive coils as shown in Figures 2-7 and 2-8. When an rf power is applied to the coil, commonly referred to as the ICP source power, the time-varying current flowing through the coil creates a magnetic flux along the axis of the cylindrical vessel. This magnetic flux induces an electric field inside the vacuum vessel. The electrons are accelerated and collide with the neutral operating gas, causing the gas molecules to be ionized, excited or fragmented, forming high-density plasma. The electrons in circular path are confined and only have a small chance of being lost to the chamber walls, thus the dc self-bias remains low. The plasma generated as described above consists of two kinds of active species, neutrals and ions. The material to be etched sits on top of a small electrode that acts as parallel plate capacitor along with the chamber as the second electrode. When an rf power, also known as electrode...
power or chuck power, is applied to the sample stage, the electrons in the plasma accelerate back and forth in the plasma from the changes in the sinusoidal field. Since electrons have much lighter mass compared to the other species in the plasma, they respond more rapidly to the frequency change than the other species. As the electrons impinge the chamber surfaces, the chamber becomes slightly negative relative to the plasma. The surface area of the chamber is larger than the sample stage, thus the negative charge is concentrated on the sample stage. This bias attracts the ions toward the sample, bombarding the surface to remove material. In an ICP system, the plasma density and the ion energy and are effectively decoupled in order to achieve uniform density and energy distributions and maintain low ion and electron energy low. This enables ICP etching to reduce plasma damage while achieving fast etch rates.

The plasma generated as described above consists of two kinds of active species: neutrals and ions. Neutrals are chemically reactive and etch the material by chemical reactions, while ions are usually less reactive and are responsible for removing material by physically bombarding the sample surface. The kinetic energy of the ions is controlled by electrode bias. The electron density and ion density are equal on average, but the density of neutrals, known as the plasma density, is typically higher. Anisotropic profiles are obtained by superimposing an rf bias on the sample to independently control ion energy and by using low pressure conditions to minimize ion scattering and lateral etching.

The plasma is neutral but is positive relative to the electrode. It appears to glow due the ion excitation from the electron movements. The recombination of charges at the boundary surfaces surrounding the plasma creates a charge depletion layer, also known as a sheath, dark space or dark region, resulting in diffusion of carriers to the boundaries. The diffusion of electrons is faster than ions initially, thus an excess of positive ions is left in the plasma and assumes a
plasma potential, $V_p$, with respect to the grounded walls. The plasma and substrate potentials generate drift current to enhance the ion motions and hinder the electron motions until steady state condition is achieved. The difference in electron and ion mobility also generates a sheath near the powered electrode. The dark region, a small region in the plasma immediately above the sample, keeps the electrons away due to the negatively charged electrode. The powered electrode reaches a self-bias negative voltage, $V_{dc}$, with respect to the ground. Even though the voltage drop controls the ion bombardment energy across the plasma sheath, it is difficult to measure; therefore, it is common to monitor the $V_{dc}$. Note that the dc bias is not a basic parameter and is characteristic to a particular piece of equipment.

Etching is accomplished by the interaction of the plasma to the substrate. The three basic etching mechanisms, chemical etch process, physical etch process, and a combination of both chemical and physical etching process, are shown in Figures 2-9, 2-10, and 2-11, respectively. Chemical etch process is the chemical reaction that etches the substrate when active species (neutrals) from the gas phase are absorbed on the surface material and react with it to form a volatile product. The chemical etch rate is limited by the chemical reaction rate or diffusion rate that depends on the volatility of the etch products since undesorbed products coat the surface and prevent or hinder further reactions. Chemical etching is a purely chemical process therefore etches isotropically, or equally in all directions. Physical process, also known as sputtering, occurs when positive ions impinge normal to the substrate surface. If the ions have sufficiently high energy, atoms, molecules or ions are ejected from the substrate surface to achieve a vertical etch profile. The etch rate of sputtering is slow, and the surface is often damaged from the ion bombardment. A combination of both chemical and physical etching process, also known as energy-driven, ion-enhanced mechanism, takes advantage of the effect of ion bombardment in
the presence of reactive neutral species. The energetic ions damage the surface and leave the surface more reactive toward incident neutrals, leading to removal rates that exceed the sum of separate sputtering and chemical etching. This process produces very fast etch rates and anisotropic profile; therefore, it is desirable in high fidelity pattern transfer.

**Ion Implantation**

Ion implantation is a physical process that introduces dopants by means of high-voltage bombardment to achieve desired electrical properties in defined areas with minimal lateral diffusion. Inside a vacuum chamber, a filament is heated to a sufficiently high temperature where electrons are created from the filament surface. The negatively charged electrons are attracted to an oppositely charged anode in the chamber. As the electrons travel from the filament to the anode, they collide and create positively charged ions from the dopant source molecules. The ions are separated in a mass analyzer, a magnetic field that allows the passage of the desired species of positive ions with specific characteristic arc radius based upon ion mass. The selected ions are accelerated in an acceleration tube and then focused into a small diameter or several parallel beams. The beam is scanned onto the wafer surface, and the ions physically bombard the wafer. The ions enter the surface and come to rest below the surface as they lose their energy through nuclear interactions and coulombic interactions, resulting in Gaussian distribution concentration profile. A schematic of an ion implantation system is illustrated in Figure 2-12.

During implantation, the collisions with high-energy ions cause crystal damage to the wafer, leading to poor electrical characteristics. In most cases, the carrier lifetime and mobility decrease drastically. Also, only a small fraction of the implanted ions are located in substitutional sites and contribute to carrier concentration. Annealing is needed to repair the crystal damage and to activate the dopants. To determine the depth and damage profile, Rutherford
Backscattering and Channeling (RBS/C) analytical technique is employed. Annealing process and RBS/C will be further discussed in the subsequent sections.

**Rapid Thermal Annealing**

Annealing is a thermal process used for repairing the ion implantation damage, diffusing dopants and alloying metal contacts. After ion implantation, annealing is employed to repair the crystal damages caused by the high-energy ion bombardment that degrade carrier lifetime and mobility. Since the majority of the implanted dopants reside in the interstitial sites, the as-implanted materials have poor electrical properties. Annealing provides thermal energy for the dopants to migrate to the substitutional sites and contribute to the carrier concentration.

Traditionally, tube furnaces were used for annealing after ion implantation. However, furnace annealing causes the implanted atoms to diffuse laterally and requires relatively long anneal time. Rapid thermal annealing was developed in order to overcome these drawbacks.

Rapid thermal annealing (RTA) utilizes radiation heating from arc lamps or tungsten-halogen lamps to heat the wafer in an inert atmosphere such as N₂ or Ar. It can attain higher temperature at a shorter time period than a conventional tube furnace, and the overall anneal time is relatively short, usually taking seconds as compared to several minutes to hours in a conventional tube furnace. RTA allows uniform heating and cooling that reduces thermal gradients that can lead to warping and stress-induced defects, enabling more dense design and fewer failures due to dislocations.

**Characterization Techniques**

**Atomic Force Microscopy**

Atomic force microscopy (AFM) employs a microscopic tip on a cantilever that deflects a laser beam depending on surface morphology and properties through an interaction between the tip and the surface. The signal is measured with a photodetector, amplified and converted into an
image display on a cathode ray tube. Depending on the type of surface, AFM can be performed in contact mode and tapping mode. A schematic diagram of AFM is shown in Figure 2-13. **Auger Electron Spectroscopy**

Auger electron spectroscopy (AES) determines the elemental composition of the few outermost atomic layers of materials. A focused beam of electrons with energies from 3 keV to 30 keV bombards the surface of a specimen. The core-level electrons are ejected from approximately 1 μm within the sample, resulting in a vacancy in the core-level. As the atom relaxes, an outer-level electron fills the core vacancy and releases excess energy, which in turn, ejects an outer electron, known as an Auger electron. This process is illustrated in Figure 2-14. The kinetic energy of the Auger electrons is characteristic of each element, with the exception of hydrogen and helium. Therefore, by measuring the energies of the Auger electrons, the near-surface composition of a specimen can be identified. In addition, AES can provide compositional depth profile from relative intensities of the elements present if the system is equipped with an ion gun to sputter away material.

**X-ray Photoelectron Spectroscopy**

X-ray photoelectron spectroscopy (XPS), also known as electron spectroscopy for chemical analysis (ESCA), provides similar information as AES. Instead of impinging the sample surface with an electron beam, XPS utilizes a monoenergetic x-ray beam to cause electrons to be ejected, usually two to 20 atomic layers deep. The variation of the kinetic energies of the ejected electrons identifies the elements present and chemical states of the elements.

**Electrical Measurements**

Current-voltage (I-V) measurements were taken to characterize the electrical properties of the contacts. These measurements are performed on an Agilent 4156 Semiconductor Parameter
Analyzer connected to a micromanipulator probe station. For diodes, the input voltage was applied through Schottky and out through ohmic for forward bias and vice versa for reverse bias measurements. For ohmic measurements on TLM pads 4 probes were used in series, two outer probes for applying the current and inner two probes for picking up the voltage.

**Photoluminescence (PL)**

Photoluminescence (PL) is an analytical technique that provides information about the optical properties of a substrate. A light source, such as He-Cd, Ar and Kr lasers, with energy larger than the bandgap energy of the semiconductor being studied, generates electron-hole pairs within the semiconductor. The excess carriers can recombine via radiative and non-radiative recombinations. Photoluminescence, the light emitted from radiative recombination, is detected. The wavelength associated with the different recombination mechanism is measured.

The luminescence from excitons, electrons and holes bound to each other, is observed only at low temperatures in highly pure materials. As the temperature increases, the exciton breaks up into free carriers from the thermal energy. Increase in doping also causes the dissociation of excitons under local electric fields. Under these conditions, the electrons and holes recombine via the band-to-band process. Since some of the electrons may not lie at the bottom of the conduction band, their recombination and holes will produce a high-energy tail in the luminescence spectrum. On the other hand, the band-to-band recombination will yield a sharp cutoff at the wavelength corresponding to the band gap of the material [118, 120].

**Rutherford Backscattering Spectrometry/Channeling**

Depth profile of implanted ions and damages can be obtained by the Rutherford Backscattering Spectrometry/Channeling (RBS/C) technique, which measures the energy distribution of the backscattered ions from the implanted sample surface at a specific angle. The
energy of the backscattered ion is determined by the mass of the atomic nucleus and the depth at which the elastic collisions take place.

A beam of high-energy ions impacts the surface of the specimen. The angle of the analyzing ions affects the penetration depth. If the ions are injected parallel to the crystal axis of the specimen, they penetrate considerably deeper than if injected randomly, due to the lower stopping power from channeling. Deeper penetration results in higher backscattered ions yield. The displacement of an atom, either as host or impurities, from the crystal lattice also increases the backscattering yield. Therefore, the distribution of displaced atoms that are caused by the radiation damage from ion implantation can be measured by increasing the backscattered ion yield\textsuperscript{118, 120}.

**Scanning Electron Microscopy**

Scanning electron microscopy (SEM) generates images from electrons instead of light. A beam of electron is produced and accelerated from an electron gun. The electron beam passes through a series of condenser and objective lenses, which focus the electron beam. A scanning coil moves the beam across the specimen surface. The electron beam interacts with the specimen, and electrons from the surface interaction volume, such as backscattered, secondary, characteristic x-ray continuous x-ray, and Auger, are emitted. The signals are collected, amplified and converted to a cathode ray tube image. Depending on the specimen and the equipment setup, the contrast in the final image provides information on the specimen composition, topography and morphology. The main advantages of using electrons for image formation are high magnification, high resolution and large depth of fields\textsuperscript{118, 120}.

**Secondary Ion Mass Spectrometry**

Secondary ion mass spectrometry (SIMS) is a highly sensitive chemical characterization technique. Primary ions, such as Cs\textsuperscript{+}, O\textsubscript{2}\textsuperscript{+}, O\textsuperscript{−} and Ar, bombard the specimen in an ultra high
vacuum environment, sputtering away secondary ions from the specimen surface. A small fraction of the ejected atoms are ionized either positively or negatively, and they are called secondary electrons. The composition of the surface is determined by the secondary electrons that are individually detected and tabulated using a mass spectrometer, as a function of their mass-to-charge ratio.

There are two modes of SIMS, static or dynamic. In the static mode, a low primary-ion flux \(<10^{14} \text{ cm}^{-2}\) is used, leaving the specimen surface relatively undisturbed. The majority of secondary ions originate in the top one or two monolayers of the samples. The dynamic mode monitors the selected secondary ion intensities as a function of the sputtering time, resulting in a concentration versus depth profile. The depth resolution of this technique ranges from 5 to 20 nm.

**Stylus Profilometry**

Stylus profilometry is used to measure the topographical features of a specimen surface, such as roughness, step height, width and spacing. A probe, or stylus, contacts the surface of the specimen and follows height variation as it scans across the surface. The height variations are converted into electrical signals, providing a cross-sectional topographical profile of the specimen. In this work, the etch rate was calculated by the depth, as measured by the profilometer, over a specified period of time.

<table>
<thead>
<tr>
<th>Property</th>
<th>Si</th>
<th>GaAs</th>
<th>GaN</th>
</tr>
</thead>
<tbody>
<tr>
<td>Bandgap energy (eV)</td>
<td>1.1 (indirect)</td>
<td>1.4 (direct)</td>
<td>3.4 (direct)</td>
</tr>
<tr>
<td>Electron mobility (cm(^2/V)s)</td>
<td>1400</td>
<td>8500</td>
<td>1000 (bulk)</td>
</tr>
<tr>
<td>Hole mobility (cm(^2/V)s)</td>
<td>600</td>
<td>400</td>
<td>2000 (2D-gas)</td>
</tr>
<tr>
<td>Electron effective mass (light)</td>
<td>0.98</td>
<td>0.067</td>
<td>0.19</td>
</tr>
<tr>
<td>Hole effective mass (light)</td>
<td>0.16</td>
<td>0.082</td>
<td>0.60</td>
</tr>
</tbody>
</table>
Table 2-2. The physical parameters in different semiconductor materials

<table>
<thead>
<tr>
<th></th>
<th>Si</th>
<th>GaAs</th>
<th>GaN</th>
<th>AlN</th>
<th>6H-SiC</th>
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<tbody>
<tr>
<td>Bandgap (eV)</td>
<td></td>
<td></td>
<td>3.4</td>
<td></td>
<td>2.9</td>
</tr>
<tr>
<td>@300°C</td>
<td>1.1</td>
<td>1.4</td>
<td></td>
<td>6.2</td>
<td></td>
</tr>
<tr>
<td>Electron mobility (cm$^2$/V-s), RT</td>
<td></td>
<td></td>
<td>1000 (bulk)</td>
<td></td>
<td>600</td>
</tr>
<tr>
<td></td>
<td>1400</td>
<td>8500</td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>Hole mobility (cm$^2$/V-s), RT</td>
<td>600</td>
<td>400</td>
<td>30</td>
<td>14</td>
<td>40</td>
</tr>
<tr>
<td>Saturation velocity (cm/s), $10^7$</td>
<td>1</td>
<td>2</td>
<td>2.5</td>
<td>1.4</td>
<td>2</td>
</tr>
<tr>
<td>Breakdown field (V/cm) x $10^6$</td>
<td>0.3</td>
<td>0.4</td>
<td>&gt;5</td>
<td></td>
<td>4</td>
</tr>
<tr>
<td>Thermal conductivity (W/cm)</td>
<td>1.5</td>
<td>0.5</td>
<td>1.5</td>
<td>2</td>
<td>5</td>
</tr>
<tr>
<td>Melting temperature (K)</td>
<td>1690</td>
<td>1510</td>
<td>&gt;1700</td>
<td>3000</td>
<td>&gt;2100</td>
</tr>
</tbody>
</table>

Table 2-3. Ionization energy of impurities for wurtzite GaN.

<table>
<thead>
<tr>
<th>Impurities</th>
<th>Ga-site (eV)</th>
<th>N-site (eV)</th>
<th>Remarks</th>
</tr>
</thead>
<tbody>
<tr>
<td>Si</td>
<td>0.012-0.02</td>
<td>0.03</td>
<td>Donor</td>
</tr>
<tr>
<td>Native defect (V$_N$)</td>
<td></td>
<td></td>
<td>Donor</td>
</tr>
<tr>
<td>C</td>
<td>0.11-0.14</td>
<td></td>
<td>Donor</td>
</tr>
<tr>
<td>Mg</td>
<td>0.14-0.21</td>
<td></td>
<td>Acceptor</td>
</tr>
<tr>
<td>Si</td>
<td></td>
<td>0.19</td>
<td>Acceptor</td>
</tr>
<tr>
<td>Zn</td>
<td>0.21-0.34</td>
<td></td>
<td>Acceptor</td>
</tr>
<tr>
<td>Native defect (v$_{ga}$)</td>
<td>0.14</td>
<td></td>
<td>Acceptor</td>
</tr>
<tr>
<td>Hg</td>
<td>0.41</td>
<td></td>
<td>Acceptor</td>
</tr>
</tbody>
</table>
Table 2-4. Basic physical properties of GaN.

<table>
<thead>
<tr>
<th>Property</th>
<th>Value</th>
</tr>
</thead>
<tbody>
<tr>
<td>Lattice parameters at 300 K (nm)</td>
<td>$a_0 = 0.3189$ nm, $c_0 = 0.5185$ nm</td>
</tr>
<tr>
<td>Density (g cm$^{-3}$)</td>
<td>6.095 g.cm$^{-3}$</td>
</tr>
<tr>
<td>Stable phase at 300 K</td>
<td>Wurtzite</td>
</tr>
<tr>
<td>Melting point (°C)</td>
<td>2500</td>
</tr>
<tr>
<td>Thermal conductivity (Wcm$^{-1}$K$^{-1}$)</td>
<td>1.3, 2.2±0.2</td>
</tr>
<tr>
<td>For thick, free-standing GaN</td>
<td></td>
</tr>
<tr>
<td>Linear thermal expansion coefficient</td>
<td>Along $a_0 = 5.59 \times 10^{-6}$ K$^{-1}$</td>
</tr>
<tr>
<td>Static dielectric constant</td>
<td>8.9</td>
</tr>
<tr>
<td>Refractive index</td>
<td>2.67 at 3.38 eV</td>
</tr>
<tr>
<td>Energy bandgap (eV)</td>
<td>Direct, 3.45</td>
</tr>
<tr>
<td>Exciton binding energy (meV)</td>
<td>26</td>
</tr>
<tr>
<td>Electron effective mass</td>
<td>0.20</td>
</tr>
</tbody>
</table>

Table 2-5. Metal work function and ideal barrier heights for GaN (electron affinity: 4.1 eV)

<table>
<thead>
<tr>
<th>Element</th>
<th>Work Function (eV)</th>
<th>Ideal Barrier Height (eV)</th>
</tr>
</thead>
<tbody>
<tr>
<td>B</td>
<td>4.45</td>
<td>0.35</td>
</tr>
<tr>
<td>Cr</td>
<td>4.5</td>
<td>0.4</td>
</tr>
<tr>
<td>Pt</td>
<td>5.64</td>
<td>1.54</td>
</tr>
<tr>
<td>Ti</td>
<td>4.33</td>
<td>0.23</td>
</tr>
<tr>
<td>W</td>
<td>4.55</td>
<td>0.45</td>
</tr>
<tr>
<td>Zr</td>
<td>4.05</td>
<td>-0.05</td>
</tr>
</tbody>
</table>
Figure 2-1. Crystal structure of wurtzite GaN.

Figure 2-2. The III-V compound semiconductor tree
Figure 2-3. Structure of a HEMT
Figure 2-4. Previous study of schottky contacts A) Index of interface behavior $S$ as a function of the electronegativity difference of the semiconductors. B) Barrier height versus work function of metals deposited on n-GaN reported from various groups.
Figure 2-5. Lithography pattern for Schottky diode

Figure 2-6. Lithography pattern for linear TLM A) TLM pads. B) Plot for measurement.
Figure 2-7. An ICP reactor.

Figure 2-8. Electric and magnetic fields inside the reactor.

Figure 2-10. Physical etching process. A) Generation of reactive species. B) Acceleration of ions to the surface. C) Ions bombard surface D) Surface atoms are ejected from the surface.

Figure 2-12. Ion implantation system.
Figure 2-13. Simplified principle of AFM.
CHAPTER 3
TUNGSTEN AND ZIRCONIUM BORIDE BASED OHMIC CONTACTS TO N-GaN

Introduction

There is a strong interest in the development of more reliable and thermally stable ohmic contacts on GaN-based electronic devices such as high electron mobility transistors (HEMTs)\textsuperscript{13-45}, which show outstanding potential in advanced power amplifiers for radar and communication systems over a broad frequency range from S-band to V-band\textsuperscript{13-15}. A key aspect of operation of nitride-based HEMTs at high powers is the need for temperature-stable high quality ohmic contacts. There is increasing interest in the application of AlGaN/GaN HEMTs to microwave power amplifiers capable of uncooled or high temperature operation where thermal stability of the contact metallurgy is paramount. The most common ohmic metallization for AlGaN/GaN HEMTs is based on Ti/Al. This bilayer must be deposited with over-layers of Ni, Ti or Pt, followed by Au to reduce sheet resistance and decrease oxidation during the high temperature anneal needed to achieve the lowest specific contact resistivity\textsuperscript{16-36}. For improving the thermal stability of ohmic contacts, there is interest in higher melting temperature metals, including W\textsuperscript{37,38,40,45}, WSi\textsubscript{x}\textsuperscript{37-39}, Mo\textsuperscript{16}, V\textsuperscript{39} and Ir\textsuperscript{40,42,43}. Another promising metallization system as the diffusion barrier layer is based on borides of Cr, Zr, Hf, Ti or W\textsuperscript{46-47}. Stoichiometric diborides have high melting temperatures (e.g. 3200 °C for ZrB\textsubscript{2}) and thermodynamic stabilities at least as good as comparable nitrides or silicides\textsuperscript{48}. These metallization systems are expected to be less reactive with GaN than the conventional Ti/Al. Previous work has shown good contact resistance obtained with Ti/Al/Mo/Au, Ti/Al/Ir/Au, Ti/Al/Pt/WSi/Ti/Au and Ti/Al/Pt/W/Ti/Au on n-GaN\textsuperscript{16,39}, but there is a broad range of other contact metallurgies that are promising, including those based on borides. For example, one attractive option is ZrB\textsubscript{2}, which has a low resistivity in the range 7-10 \(\mu\Omega.cm\). Recently, it was shown that hexagonal ZrB\textsubscript{2} (0001) single crystals have an in-
plane lattice constant close to that of GaN, prompting efforts at GaN heteroepitaxy on ZrB$_2$ or buffer layers on Si substrates. For contacts on n-type GaN, the only related work is the study of ZrN/Zr/n-GaN Ohmic structures $^{48}$ in which the Zr/GaN interface was found to have excellent thermal stability. Even though we are not relying on the ZrB$_2$ to make direct ohmic contact to GaN, it is expected that ZrB$_2$ contacts on GaN will have low barrier heights, given the work function of ZrB$_2$ is $\sim$3.94 eV and the electron affinity of GaN is $\sim$4.1 eV.

In this chapter a report on the annealing temperature dependence of contact resistance and contact intermixing of Ti/Al/W$_2$B/Ti/Au and of Ti/Al/ZrB$_2$/Ti/Au on n-GaN is given. The Tungsten Boride contacts show excellent minimum contact resistance of $7 \times 10^{-6}$ $\Omega$.cm$^2$ after annealing at 800 °C and promising long-term stability at 200 °C. The ZrB$_2$ based contacts show excellent minimum contact resistance of $3 \times 10^{-6}$ $\Omega$.cm$^2$ after annealing at 700 °C and retain a good morphology even after annealing at 1000 °C.

**Experimental**

The samples used were 3 $\mu$m thick Si-doped GaN grown by Metal Organic Chemical Vapor Deposition on c-plane Al$_2$O$_3$ substrates. The electron concentration obtained from Hall measurements was $\sim$7x$10^{18}$ cm$^{-3}$. Mesas 1.8 $\mu$m deep were formed by Cl$_2$/Ar Inductively Coupled Plasma Etching to provide electrical isolation of the contact pads. A metallization scheme of Ti (200 Å)/Al (1000 Å)/ W$_2$B or ZrB$_2$ (500 Å) / Ti (200 Å) /Au (800 Å) was used in these experiments. All of the metals were deposited by Ar plasma-assisted rf sputtering at pressures of 15-40 mTorr and rf (13.56 MHz) powers of 200-250 W. The contacts were patterned by liftoff and annealed at 500-1000 °C for 1 min in a flowing N$_2$ ambient in a RTA furnace. Auger Electron Spectroscopy (AES) depth profiling of the as-deposited contacts showed sharp interfaces between the various metals in both types of contacts. For the AES analysis, the
samples were mounted on a stainless steel puck and placed in the system load-lock. Clean tweezers and gloves were used for all sample handling. No additional cleaning steps were implemented. After sufficient evacuation, the sample puck was inserted into the analytical chamber and placed in front of the analyzer. The AES system was a Physical Electronics 660 Scanning Auger Microprobe. The electron beam conditions were 10 keV, 1 µA beam current at 30° from sample normal. For depth profiling, the ion beam conditions were 3 keV Ar⁺, 2.0µA (3 mm)² raster, with sputter rate based on 110 Å/minute (SiO₂) of 142 Å/ minute Au (1.3*SiO₂),64 Å/minute Ti (0.58*SiO₂),123 Å/minute Al (1.12*SiO₂), 44 Å/minute W (0.4*SiO₂) {for W₂B},55 Å/minute Al₂O₃ (0.5*SiO₂) and 83 Å/minute (0.75*SiO₂). Prior to AES data acquisition, secondary electron images (SEI’s) were obtained from the sample. The SEIs were obtained at magnifications of 125X, and 1,000X. The SEIs were used to locate and document analysis area locations and to document surface morphology. The quantification of the elements was accomplished by using the elemental sensitivity factors. The contact properties were obtained from linear transmission line method (TLM) measurements on 100×100 µm pads with spacing 5, 10, 20, 40, and 80 µm. The contact resistance $R_C$ was obtained from the relation $R_C = (R_T \rho_S d/Z)/2$, where $R_T$ is the total resistance between two pads, $\rho_S$ is the sheet resistivity of the semiconductor under the contact, $d$ is the pad spacing, and $Z$ is the contact width. The specific contact resistance, $\rho_C$, is then obtained from $\rho_C = R_C L_T Z$, where $L_T$ is the transfer length obtained from the intercept of a plot of $R_T$ vs $d$.

**Results and Discussion**

**Tungsten Boride Based Ohmic Contact**

Figure 3-1 shows the contact resistance as a function of anneal temperature. The as-deposited contacts showed rectifying behavior. This is expected for any as-deposited contacts on the wide bandgap GaN. The current-voltage characteristics became ohmic for anneal
temperatures ≥ 500 °C. The contact resistance decreased up to ~800 °C, reaching a minimum value of 7x10^{-6} Ω.cm^2. This trend is most likely related to the formation of low resistance phases of TiN at the interface with the GaN, as reported for conventional Ti/Al/Pt/Au contacts^{19-22}. However the W_2B-based contacts show improved edge acuity, which is important for small gate length HEMTs in order to reduce the possibility of shorting of the ohmic metal to the gate. Annealing at higher temperatures leads to higher contact resistance, which as will be seen later corresponds to extensive intermixing of the contact metallurgy. The corresponding transfer resistance and semiconductor sheet resistance data are shown in Figure 3-2. The minimum contact resistance obtained corresponds to a transfer resistance of 0.057 Ω.mm.

Figure 3-3 shows the measurement temperature dependence of the Ti/Al/W_2B/Ti/Au on n-GaN annealed at 800 °C. Over the relatively limited temperature range available for the measurements and within the error of the measurement, we did not observe any temperature dependence. This indicates that at this anneal temperature, the dominant current transport mechanism is field emission^{49}, since thermionic emission would have significant temperature dependence and thermionic field emission is operative at lower doping ranges (10^{16}-10^{18} cm^{-3}).

Figure 3-4 shows the SEI of the as-deposited contact morphology and after annealing at 500, 800 or 1000 °C. The morphology is featureless until 800 °C, which corresponds to the minimum in contact resistance. By 1000 °C, the morphology becomes very rough and this corresponds to the increased contact resistance.

The AES surface scans as a function of anneal temperature was also obtained. Carbon, oxygen and gold were detected on the as-deposited surface. The carbon is adventitious and the oxygen comes from a thin native oxide on the Au. After annealing at 500 °C, titanium was detected on the contact surface. After 800 °C annealing, gold, titanium, aluminum, and gallium
were detected on the surface, which is consistent with the onset of extensive reaction of the contact metallurgy. After 1000 °C annealing, titanium, aluminum, and gallium were detected on the surface. The surface concentration of gold decreased with increasing temperature of the annealing step. Titanium and aluminum concentrations increased with annealing temperature.

Figure 3-5 shows the AES depth profiles for the as-deposited and annealed samples. The profile obtained from the as-deposited sample was in good agreement with the prescribed metal layer thicknesses. Oxygen was detected at the Ti/W₂B interface, in the W₂B layer, at the W₂B/Al interface, and in the deep Ti layer. Note that the main (and only) boron peak overlaps with one of the tungsten peaks. The overlap results in an overestimation of boron in the W₂B film. Also note that nitrogen was not plotted due to a peak overlap with titanium however nitrogen should only be present in the GaN substrate.

The profile obtained from the sample annealed at 500 °C shows diffusion of titanium through the gold layer to the surface. The concentration of oxygen in the titanium layers is higher in this sample compared to the as deposited sample although the distributions of oxygen throughout the profiles are similar. The profile obtained from the sample annealed at 800 °C shows significant diffusion or inter-diffusion of layers. The profile now shows the presence of a thin titanium layer, followed by a thin gold layer, then an oxidized aluminum layer, the W₂B layer, another oxidized aluminum layer, a gold layer, and a final titanium layer. Layer thickness values should be considered approximate. The sputter rates are only appropriate for pure elements or compounds. Diffusion of other species into any given layer will probably have an impact on the sputter rate of that material. The profile obtained from the sample annealed at 1000 °C shows titanium at the surface, followed by an oxidized aluminum layer, a gold layer, the W₂B layer, another gold layer, a titanium layer, and the GaN substrate. Some of the transport might be
attributed to grain-boundary diffusion, as reported previously for Mo-based contacts to AlGaN/GaN heterostructures \(^{39}\).

Figure 3-6 shows the room temperature contact resistance of the sample annealed at 800 °C, as a function of time spent at 200 °C. This simulates the operation of an uncooled GaN-based transistor and gives some idea of the expected stability of the contact. Within experimental error, we did not observe any degradation of contact resistance over a period of almost 3 weeks. Future work will establish the stability of the metallization over longer periods and at higher temperatures.

**Zirconium Boride Based Ohmic Contacts**

Figure 3-7 A shows the contact resistance as a function of annealing temperature, while the associated GaN sheet resistance under the contact is shown at the bottom of the figure. The as-deposited contacts were rectifying, with a transition to ohmic behavior for anneal temperatures \(\geq 500 \, ^{\circ}C\). The contact resistance decreased up to \(~700 \, ^{\circ}C\), reaching a minimum value of \(3\times10^{-6} \, \Omega\cdot\text{cm}^2\). This same basic trend is seen in most Ti/Al-based contacts \(^{19-22}\) and is attributed to formation of low resistance phases of TiN at the interface with the GaN. By comparison with the usual Ti/Al/Pt/Au metal stack, the ZrB\(_2\)-based contacts show improved edge acuity, an important factor for small gate length HEMTs in order to reduce the possibility of shorting of the ohmic metal to the gate. The ZrB\(_2\)-based contacts show a double minimum in contact resistance versus annealing temperature and even at 1000 °C exhibit a contact resistance below \(10^{-5} \, \Omega\cdot\text{cm}^2\).

Figure 3-8 shows the specific contact resistivity (A) and sheet resistance (B) as a function of anneal time at 700 °C for Ti/Al/ZrB\(_2\)/Ti/Au on n-GaN. The minimum contact resistance is achieved for 60 s anneals. This is also consistent with the need to form a low resistance interfacial phase whose formation kinetics is probably limited by diffusion of Ti to the GaN interface. Figure 3-9 shows the measurement temperature dependence of the Ti/Al/ZrB\(_2\)/Ti/Au
on n-GaN annealed at 700 °C. We did not observe any significant temperature dependence, suggesting that at this anneal temperature the dominant current transport mechanism is field emission\textsuperscript{49}.

Figure 3-10 shows scanning electron microscopy (SEM) images of the as-deposited contact morphology and after annealing at 500, 700 or 1000 °C. Even at the highest anneal temperature, the morphology remains quite smooth on the scale accessible to the SEM. This is in sharp contrast to the case of Ti/Al/Pt/Au, where significant roughening occurs above 800 °C and this result suggests that the ZrB\textsubscript{2} is an effective barrier for reducing intermixing of the contact compared to Pt.

Figure 3-11 shows the AES surface scans as a function of anneal temperature. As expected, only carbon, oxygen and gold were detected on the as-deposited surface. The carbon is adventitious and the oxygen comes from a thin native oxide on the Au. After annealing at 500 °C, titanium and aluminum was detected on the contact surface and their concentrations increased at higher annealing temperatures. The surface concentration of gold decreased with increasing temperature of the annealing step. After 1000 °C annealing, Boron was also present on the surface, having outdiffused from the buried ZrB\textsubscript{2} layer. Table 3-1 summarizes the near-surface composition data obtained from AES measurements.

Figure 3-12 shows the AES depth profiles for the as-deposited and annealed samples. The profile obtained from the as-deposited sample was in good agreement with the prescribed metal layer thicknesses. Oxygen was detected in the ZrB\textsubscript{2} layer, consistent with past observations that the borides are excellent getters for water vapor during deposition\textsuperscript{47}. The profile obtained from the sample annealed at 500 °C shows extensive diffusion of titanium through the gold layer to the surface. The intermixing of the contact metallurgy becomes more pronounced as the
annealing temperature is increased. Some of the transport might be attributed to grain-boundary diffusion, as reported previously for Mo-based contacts to AlGaN/GaN heterostructures. As noted earlier, the contact morphology does not degrade significantly even after 1000 °C annealing.

Summary and Conclusions

Both Ti/Al/W₂B/Ti/Au and Ti/Al/ZrB₂/Ti/Au metallization scheme were used to form ohmic contacts to n-type GaN. For Tungsten Boride based contact, a minimum specific contact resistivity of 7x10⁻⁶ Ω.cm² was achieved at an annealing temperature of 800 °C, which is comparable to that achieved with conventional Ti/Al/Pt/Au on the same samples. For Zirconium Boride based contact, a minimum specific contact resistivity of 3x10⁻⁶ Ω.cm² was achieved at an annealing temperature of 700 °C, which is comparable to that achieved with conventional Ti/Al/Pt/Au on the same samples. The ZrB₂-based contact appears to have greater thermal stability than the conventional metallization. The Tungsten Boride based contacts showed no change in resistance over a period of more than 450 hours at 200 °C. This approach of using boride-based contacts looks promising for high temperature device applications.

Table 3-1. Near-surface composition of contact stack determined by AES measurements for ZrB₂ ohmic contact

<table>
<thead>
<tr>
<th>Sample ID</th>
<th>C(1)</th>
<th>O(1)</th>
<th>Al(1)</th>
<th>S(1)</th>
<th>Ti(2)</th>
<th>B(2)</th>
<th>Au(3)</th>
</tr>
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<tbody>
<tr>
<td>Sensitivity factors</td>
<td>[0.076]</td>
<td>[0.212]</td>
<td>[0.000]</td>
<td>[0.652]</td>
<td>[0.000]</td>
<td>[0.000]</td>
<td>[0.049]</td>
</tr>
<tr>
<td>#1: As deposited</td>
<td>48</td>
<td>1</td>
<td>Nd</td>
<td>&lt;1</td>
<td>3</td>
<td>nd</td>
<td>49</td>
</tr>
<tr>
<td>#2: Annealed at 500 °C</td>
<td>41</td>
<td>22</td>
<td>16</td>
<td>&lt;1</td>
<td>3</td>
<td>nd</td>
<td>17</td>
</tr>
<tr>
<td>#3: Annealed at 700 °C</td>
<td>36</td>
<td>29</td>
<td>22</td>
<td>&lt;1</td>
<td>5</td>
<td>nd</td>
<td>8</td>
</tr>
<tr>
<td>#4: Annealed at 1000 °C</td>
<td>28</td>
<td>33</td>
<td>18</td>
<td>1</td>
<td>4</td>
<td>15</td>
<td>Nd</td>
</tr>
</tbody>
</table>
Figure 3-1. Specific contact resistivity versus anneal temperature for Ti/Al/W₂B/Ti/Au on n-GaN.
Figure 3-2. Measurement as a function of annealing temperature Ti/Al/W₂B/Ti/Au on n-GaN. A) Transfer resistance. B) Sheet resistance.
Figure 3-3. Specific contact resistance versus measurement temperature for Ti/Al/W$_2$B/Ti/Au on n-GaN annealed at 800 °C.
Figure 3-4. Secondary electron images of the Ti/Al/W_2B/Ti/Au contacts on n-GaN. A) As-deposited. B) Annealed at 500 °C. C) Annealed at 800 °C. D) Annealed at 1000 °C.
Figure 3-5. AES depth profiles of the Ti/Al/W₂B/Ti/Au on n-GaN. A) As-deposited. B) Annealed at 500 °C. C) Annealed at 800 °C. D) Annealed at 1000 °C.
Figure 3-6. Contact resistance of the Ti/Al/W₂B/Ti/Au on n-GaN, initially annealed at 800 °C, as a function of subsequent time at 200 °C.
Figure 3-7. Measurement versus anneal temperature for Ti/Al/ZrB₂/Ti/Au on n-GaN. A) Specific contact resistivity. B) Sheet resistance
Figure 3-8. Measurement as a function of annealing time at 700 °C for Ti/Al/ZrB$_2$/Ti/Au on n-GaN. A) Specific contact resistivity. B) Sheet resistance.
Figure 3-9. Specific contact resistance versus measurement temperature for Ti/Al/ZrB$_2$/Ti/Au on n-GaN annealed at 800 °C.
Figure 3-10. Secondary electron images of the Ti/Al/ZrB$_2$/Ti/Au on n-GaN. A) As-deposited. B) Annealed at 500 °C. C) Annealed at 700 °C. D) Annealed at 1000 °C.
Figure 3-11. AES surface scans of the Ti/Al/ZrB$_2$/Ti/Au on n-GaN. A) As-deposited. B) Annealed at 500 °C. C) Annealed at 700 °C. D) Annealed at 1000 °C.
Figure 3-12. AES depth profiles of the Ti/Al/ZrB₂/Ti/Au on n-GaN. A) As-deposited. B) Annealed at 500 °C. C) Annealed at 700 °C. D) Annealed at 1000 °C.
CHAPTER 4
COMPARISON OF ELECTRICAL AND RELIABILITY PERFORMANCE OF TiB$_2$, CrB$_2$
AND W$_2$B$_5$ BASED OHMIC CONTACTS ON N-GaN

Introduction

One of the remaining obstacles to commercialization of GaN high electron mobility transistors (HEMTs) power amplifiers is the development of more reliable and thermally stable ohmic contacts$^{13,15,20-64}$. These power amplifiers show great potential for radar and communication systems over a broad frequency range from S-band to V-band$^{13-15,50}$. GaN-based HEMTs can operate at significantly higher power densities and higher impedance than currently used GaAs devices$^{15,51-63}$. A key aspect of operation of nitride-based HEMTs at high powers is the need for temperature-stable high quality ohmic contacts$^{20-30,32,33,35-45,64}$. The most common ohmic metallization for AlGaN/GaN HEMTs is based on Ti/Al. This bilayer must be deposited with over-layers of Ni, Ti or Pt, followed by Au to reduce sheet resistance and decrease oxidation during the high temperature anneal needed to achieve the lowest specific contact resistivity$^{20-30,32,33,35-45,64}$. There is a lateral flow issue with these contacts due to the low melting temperature viscous AlAu$_4$ phase that may cause problems when the gate/source contact separation is small. For improving the thermal stability of ohmic contacts, there is interest in high temperature metals such as W$^{37,38,40,45}$, WSi$_x$$^{37-39}$, Mo$^{16}$, V$^{39}$ and Ir$^{40,42,43}$. An unexplored class of potentially stable contacts is that of boride-based contacts such as CrB$_2$, TiB$_2$ and W$_2$B$_5$. Some of the properties of these metals are shown in Table 4-1. They have high melting temperatures, good electrical conductivity, the heat of formation for stoichiometric borides is comparable to silicides or nitrides$^{47}$ and although there is a lack of information on phase diagrams with GaN, these metals have shown corrosion resistance against molten metals and thus should exhibit even less solubility in the solid state.
In this chapter a report on the initial results for annealing temperature dependence of contact resistance and contact intermixing of Ti/Al/boride/Ti/Au on n-type GaN, with W$_2$B$_5$, TiB$_2$ and CrB$_2$ as the three different selected borides is presented. These contacts show promising long-term stability at 350 °C which is very important for devices which are going to be used in uncooled and/or prolonged heated environment.

**Experimental**

The samples used were 3 μm thick Si-doped GaN grown by Metal Organic Chemical Vapor Deposition on c-plane Al$_2$O$_3$ substrates. The electron concentration obtained from Hall measurements was ~7x10$^{18}$ cm$^{-3}$. Mesas 1.8 μm deep were formed by Cl$_2$/Ar Inductively Coupled Plasma Etching to provide electrical isolation of the contact pads. A metallization scheme of Ti (200 Å)/Al (1000 Å)/ Boride (500 Å) / Ti (200 Å) /Au (800 Å) was used in these experiments where Borides were W$_2$B$_5$, CrB$_2$ and TiB$_2$. All of the metals were deposited by Ar plasma-assisted rf sputtering at pressures of 15-40 mTorr and rf (13.56 MHz) powers of 200-250 W. The contacts were patterned by liftoff and annealed at 500-1000 °C for 1 min in a flowing N$_2$ ambient in a RTA furnace.

Auger Electron Spectroscopy (AES) depth profiling of the as-deposited contacts showed sharp interfaces between the various metals in both types of contacts. The AES system was a Physical Electronics 660 Scanning Auger Microprobe. The electron beam conditions were 10 keV, 1 μA beam current at 30° from sample normal. For depth profiling, the ion beam conditions were 3 keV Ar$^+$, 2.0μA (3 mm)$^2$ raster. Prior to AES data acquisition, secondary electron microscopy images (SEMs) were obtained from the sample. The SEMs were obtained at magnifications of 125X, and 1,000X. The SEMs were used to locate and document analysis area
locations and to document surface morphology. The quantification of the elements was accomplished by using the elemental sensitivity factors.

The contact properties were obtained from linear transmission line method (TLM) measurements on 100×100 µm pads with spacing 5, 10, 20, 40, and 80 µm. The contact resistance \( R_C \) was obtained from the relation \( 49 \ R_C = (R_T \rho_S d/Z)/2 \), where \( R_T \) is the total resistance between two pads, \( \rho_S \) is the sheet resistivity of the semiconductor under the contact, \( d \) is the pad spacing, and \( Z \) is the contact width. The specific contact resistance, \( \rho_C \), is then obtained from \( \rho_C = R_CL_TZ \), where \( L_T \) is the transfer length obtained from the intercept of a plot of \( R_T \) vs \( d \).

**Results and Discussion**

Figure 4-1 shows the contact resistances as a function of anneal temperature for the three-boride-based schemes. The as-deposited contacts showed rectifying behavior, as expected for any as-deposited contacts on the wide bandgap GaN. The current-voltage characteristics became Ohmic for anneal temperatures \( \geq 500 \) °C. The contact resistance decreased up to \( \sim 800-900 \) °C, depending on the boride employed. The TiB\(_2\)-containing contacts show the lowest contact resistance of \( 1.6 \times 10^{-6} \, \Omega.\text{cm}^2 \). The minimum in the contact resistance with annealing temperature is most likely related to the formation of low resistance phases of TiN at the interface with the GaN, as reported for conventional Ti/Al/Pt/Au contacts. Annealing at higher temperatures leads to higher contact resistance, which as will be seen later corresponds to extensive intermixing of the contact metallurgy. The contact properties did not show a significant dependence on annealing time at 850 °C, as shown in Figure 4-2.

Figure 4-3 shows the measurement temperature dependence of the contacts on n-GaN annealed at 850 °C. Within the error of the measurement, the contacts exhibited almost constant specific contact resistance in the temperature range of 25-200 °C, indicating that current flow is
dominated by tunneling. When the tunneling dominates, the specific contact resistivity ($R_{SCR}$) is dependent upon doping concentration and is given as

$$ R_{SCR} \propto \exp\left[ \frac{2\sqrt{\varepsilon_s m_e^*}}{\hbar} \left( \frac{\phi_B}{\sqrt{N_D}} \right) \right] $$  

(4-1)

where $\phi_B$ is the barrier height, $\varepsilon_s$ the semiconductor permittivity, $m_e^*$ the effective mass of electrons, $\hbar$ the Planck’s constant and $N_D$ is the donor concentration in the semiconductor. The tunneling may be related to the formation of the TiN$_X$ phases, as is the case with conventional Ohmic contacts on n-GaN.

Figure 4-4 shows the SEM of the contact morphology for the three metallization schemes before and after annealing at 800 or 1000 °C. The morphology is featureless until 800 °C, which corresponds to the minimum in contact resistance. By 1000 °C, the morphology becomes rougher and this corresponds to the increased contact resistance

AES surface scans showed only the presence of carbon, oxygen and gold on the as-deposited surface. The carbon is adventitious and the oxygen comes from a thin native oxide on the Au. Figure 4-5 shows the AES depth profiles for the as-deposited and annealed Ti/Al/CrB$_2$/Ti/Au samples. The profile obtained from the as-deposited sample was in good agreement with the prescribed metal layer thicknesses. The profile from the sample annealed at 700 °C shows diffusion of titanium through the gold layer to the surface and of the Ti layer near the GaN through the Al above it. The profile obtained from the sample annealed at 800 °C shows significant inter-diffusion of layers. The profile now shows the presence of a thin titanium layer, followed by a thin gold layer on top of the Cr$_2$B layer, an oxidized aluminum layer, a gold layer, and a final titanium layer. The profile obtained from the sample annealed at 1000 °C shows a similar basic structure. Some of the transport might be attributed to grain-boundary diffusion, as reported previously for Mo-based contacts to AlGaN/GaN heterostructures.$^{39}$
Similar data is shown in Figures 4-6 and 4-7 for TiB$_2$ and W$_2$B$_5$ for various anneal temperatures. Once again, the movement of the two Ti layers is the major effect present in both types of contact. The W$_2$B$_5$ is the least thermally stable of the schemes, as judged by the more extensive intermixing at lower temperatures.

Figure 4-8 shows the room temperature contact resistance of the samples annealed at 800 °C, as a function of time spent at 350 °C. This simulates the operation of an uncooled GaN-based transistor and gives some idea of the expected stability of the contact. We have also included data from conventional Ti/Al/Ni/Au contacts for comparison. Note that the latter shows the lowest initial contact resistance, but then has an increase of almost an order of magnitude after 9 days of elevated temperature operation. By sharp contrast, all of the boride-based contacts show less change with aging time and have lower contact resistances than the Ti/Al/Ni/Au after 22 days aging at 350 °C. This suggests that the improved stability of the borides relative to Ni has some beneficial effect on the long-term stability of the contacts.

**Summary and Conclusions**

A Ti/Al/X/Ti/Au metallization scheme, where X was either W$_2$B$_5$, CrB$_2$ or TiB$_2$, was used to form Ohmic contacts to n-type GaN. A minimum contact resistance of $1.5 \times 10^{-6}$ Ω.cm$^2$ was achieved for the TiB$_2$ based scheme at an annealing temperature of 850-900 °C. For W$_2$B$_5$ the minimum contact resistance was $\sim 1.5 \times 10^{-5}$ Ω.cm$^2$ at 800 °C while for CrB$_2$ it was $8 \times 10^{-6}$ Ω.cm$^2$ at 800 °C. The latter value is comparable to that achieved with conventional Ti/Al/Ni/Au on the same samples. The contacts showed much less change in resistance over a period of more than 22 days at 350 °C than for Ti/Al/Ni/Au and the boride-based contacts look promising for high temperature device applications.
<table>
<thead>
<tr>
<th>Properties</th>
<th>TiB$_2$</th>
<th>ZrB$_2$</th>
<th>W$_2$B</th>
<th>W$_2$B$_5$</th>
<th>CrB$_2$</th>
</tr>
</thead>
<tbody>
<tr>
<td>Melting Point (°C)</td>
<td>2980</td>
<td>3040</td>
<td>~2670</td>
<td>~2385</td>
<td>2200</td>
</tr>
<tr>
<td>~3225</td>
<td>~3200</td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>Structure</td>
<td>hexagonal</td>
<td>hexagonal</td>
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<td>hexagonal</td>
<td>hexagonal</td>
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<tr>
<td>Thermal expansion coefficients x 10$^6$ (/deg)</td>
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<td>5.9</td>
<td>-</td>
<td>-</td>
<td>10.5</td>
</tr>
<tr>
<td>Phonon component of heat conduction (wt/m-deg.)</td>
<td>20.6</td>
<td>18.9</td>
<td>-</td>
<td>-</td>
<td>10.4</td>
</tr>
<tr>
<td>Elastic modulus E x 10$^{-6}$ (kg/cm$^2$)</td>
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<td>4.3</td>
<td>-</td>
<td>-</td>
<td>2.5</td>
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<tr>
<td>Characteristic Temperature (°K)</td>
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<td>-</td>
<td>726</td>
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<tr>
<td>Density of electronic states g x 10$^{-21}$ (eV$^{-1}$ cm$^{-1}$)</td>
<td>4.50</td>
<td>4.76</td>
<td>-</td>
<td>-</td>
<td>54.6</td>
</tr>
<tr>
<td>Work function (eV) (approx.)</td>
<td>4.19(?)</td>
<td>3.94(?)</td>
<td>-</td>
<td>-</td>
<td>3.18(?)</td>
</tr>
<tr>
<td>Heat of Formation (Kcal/mole)</td>
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<td>76.0</td>
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<td>-</td>
<td>31.0</td>
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<td>3.169</td>
<td>-</td>
<td>2.982</td>
<td>2.969</td>
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<tr>
<td>Thermal conductivity (W.m$^{-1}$.K$^{-1}$)</td>
<td>26</td>
<td>80</td>
<td>-</td>
<td>unknown</td>
<td>32</td>
</tr>
<tr>
<td>Electrical resistivity (μOhm.cm)</td>
<td>28</td>
<td>4.6</td>
<td>-</td>
<td>19</td>
<td>21</td>
</tr>
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</table>
Figure 4-1. Specific contact resistance of Ti/Al/boride/Ti/Au ohmic contacts on n-GaN as a function of anneal temperature.
Figure 4-2. Specific contact resistance of Ti/Al/boride/Ti/Au ohmic contacts on n-GaN as a function of anneal time at the optimum anneal temperature for each type of metal scheme.
Figure 4-3. Specific contact resistance of Ti/Al/boride/Ti/Au ohmic contacts on n-GaN as a function of measurement temperature at the optimum anneal temperatures.
Figure 4-4. SEM micrographs. A) As-deposited CrB\textsubscript{2}. B) As-deposited TiB\textsubscript{2}. C) As-deposited W\textsubscript{2}B\textsubscript{5}. D) CrB annealed at 800 °C. E) TiB\textsubscript{2} annealed at 800 °C. F) W\textsubscript{2}B\textsubscript{5} annealed at 800 °C. G) CrB annealed at 1000 °C. H) TiB\textsubscript{2} annealed at 1000 °C. I) W\textsubscript{2}B\textsubscript{5} annealed at 1000 °C.
Figure 4-5. AES depth profiles of CrB\textsubscript{2}-based contacts. A) As-deposited. B) Annealed at 700 °C. C) Annealed at 800 °C. D) Annealed at 1000 °C.
Figure 4-6. AES depth profiles of TiB$_2$-based contacts. A) As-deposited. B) Annealed at 600 °C. C) Annealed at 800 °C. D) Annealed at 1000 °C.
Figure 4-7. AES depth profiles of W$_2$B$_5$-based contacts. A) As-deposited. B) Annealed at 500 °C. C) Annealed at 700 °C. D) Annealed at 1000 °C.
Figure 4-8. Specific contact resistance of the boride-based contacts annealed at 800°C and the conventional Ti/Al/Ni/Au contacts as a function of aging time at 350 °C.
CHAPTER 5
ZrB₂ AND W₂B SCHOTTKY DIODE CONTACTS ON N-GaN

Introduction

AlGaN/GaN high electron mobility transistors (HEMTs) have potential application in microwave power amplifiers for radar and communication systems over a broad frequency range from S-band to V-band. One critical requirement for commercialization of these systems is the need for more reliable and thermally stable schottky contacts on n-type GaN. The anticipated operation of these amplifiers under uncooled, high temperature conditions emphasizes that the thermal stability of the contact metallurgy is paramount. An alternative approach for the gate contact is the use of metal-oxide-semiconductor (MOS) gates, though much of that work is in its infancy and the metal gate is still the dominant technology. The HEMT device structure is relatively simple and the reliability is determined by the stability of gate and source/drain contacts and surface and buffer layer trapping effects. In GaN as in other compound semiconductor systems, the strength of interfacial reactions between the metal and semiconductor plays a key role in determining the quality of the resultant schottky barriers. The most common schottky metallization for AlGaN/GaN HEMTs is based on Pt/Au or Ni/Au.

Metallurgy systems with high melting temperatures and good thermodynamic stability such as W (eg. W, WSiₓ) show potential for improved thermal characteristics on GaN. Tungsten-based schemes have been used for both rectifying (W/Ti/Au; WSiₓ/Ti/Au) and ohmic (Ti/Al/Pt/ W/Ti/Au) contacts on GaN HEMTs. Sputter-deposited pure W schottky contacts on n-GaN show as-deposited barrier heights (φₜ) of 0.80 eV for optimized conditions. Subsequent annealing at 500-600 °C reduces the barrier height to ~0.4 eV, its theoretical value from the relation φₜ=φₘ−χₛ (where φₘ is the metal work function and χₛ the electron affinity of GaN).
Another promising metallization system is based on borides of Cr, Zr, Hf, Ti or W \textsuperscript{46}. Stoichiometric diborides have high melting temperatures (eg. 3200 °C for ZrB\textsubscript{2}) and thermodynamic stabilities at least as good as comparable nitrides or silicides \textsuperscript{47}. These have been suggested as metal gates in Si complementary metal oxide semiconductor (CMOS) integrated circuits \textsuperscript{47}. One particularly attractive option is ZrB\textsubscript{2}, which has a low resistivity in the range 7-10 \(\mu\Omega\cdot\text{cm}\). To date, there is basically no information on the contact properties of ZrB\textsubscript{2} on n-GaN and W\textsubscript{2}B which is a refractory material has also not been explored as a contact to GaN. Recently, it was shown that hexagonal ZrB\textsubscript{2} (0001) single crystals have an in-plane lattice constant close to that of GaN, prompting efforts at GaN heteroepitaxy on ZrB\textsubscript{2} \textsuperscript{69-72} or buffer layers on Si substrates \textsuperscript{73}. In terms of contacts on GaN, the only related work is the study of ZrN/Zr/n-GaN ohmic structures \textsuperscript{48} in which the Zr/GaN interface was found to have excellent thermal stability. A potential drawback for application of ZrB\textsubscript{2} contacts on GaN is that the barrier height might be lower than the more conventional schemes, given the work function of ZrB\textsubscript{2} is \(\sim3.94\ \text{eV}\) and the electron affinity of GaN is \(\sim4.1\ \text{eV}\).

In this chapter a report on the annealing temperature dependence of barrier height and contact intermixing of ZrB\textsubscript{2}/Ti/Au and W\textsubscript{2}B/Ti/Au on n-GaN is presented. The ZrB\textsubscript{2} based contacts show a maximum barrier height of 0.55 eV and maintain a barrier height \(>0.5\ \text{eV}\) to at least 700 °C and the W\textsubscript{2}B based contacts show a maximum barrier height of 0.5 eV, with a negative temperature coefficient, and are stable against annealing up to \(\sim500\ \text{°C}\).

**Experimental**

The samples used were 3 \(\mu\text{m}\) thick Si-doped GaN grown by Metal Organic Chemical Vapor Deposition on c-plane Al\textsubscript{2}O\textsubscript{3} substrates. The electron concentration obtained from Hall measurements was \(\sim5\times10^{17}\ \text{cm}^{-3}\). A schottky metallization scheme of ZrB\textsubscript{2} (500 Å) / Ti (200 Å)
/Au (800 Å) was used in all experiments. The Au was added to lower the contact sheet resistance, while the Pt is a diffusion barrier. All of the metals were deposited by Ar plasma-assisted rf sputtering at pressures of 15-40 mTorr and rf (13.56 MHz) powers of 200-250 W. The contacts were patterned by liftoff and annealed at temperatures up to 700 °C for 1 min in a flowing N₂ ambient in a RTA furnace. For ohmic contacts, we used the standard e-beam deposited Ti/Al/Pt/Au annealed at 850 °C for 30 secs prior to deposition of the schottky metallization. A ring contact geometry for the diodes was employed. Figure 5-1 A shows a scanning electron microscopy (SEM) image of the as-deposited W₂B based contacts.

Auger Electron Spectroscopy (AES) depth profiling of the as-deposited contacts showed sharp interfaces between the various metals. The AES system was a Physical Electronics 660 Scanning Auger Microprobe. The electron beam conditions were 10 keV, 1 μA beam current at 30° from sample normal. For depth profiling, the ion beam conditions were 3 keV Ar⁺, 2.0μA (3 mm)² raster, with sputter rate of~160 Å / minute. Prior to AES data acquisition, secondary electron images were obtained from the sample. These images were used to locate and document analysis area locations and to document surface morphology. The quantification of the elements was accomplished by using the elemental sensitivity factors. The contact properties were obtained from I-V characteristics of the ZrB₂/Ti/Au diodes measured at 300 K and W₂B/Ti/Au diodes measured over the temperature range 25-150 °C using a probe station and Agilent 4145B parameter analyzer. We fit the forward I-V characteristics to the relation for the thermionic emission over a barrier:

\[
J_F = A^* T^2 \exp\left(-\frac{e \phi_b}{kT}\right) \exp\left(-\frac{eV}{nkT}\right)
\]

5-1
where $J$ is the current density, $A^*$ is the Richardson’s constant for n-GaN, $T$ the absolute temperature, $e$ the electronic charge, $\phi_b$ the barrier height, $k$ Boltzmann’s constant, $n$ the ideality factor and $V$ the applied voltage.

**Results and Discussion**

**$W_2B$ Based Rectifying Contacts**

Figure 5-2 shows the extracted barrier height and reverse breakdown voltage as a function of measurement temperature for as-deposited $W_2B/Ti/Au$ contacts on n-GaN. From the data, $\phi_b$ was obtained as 0.55 eV for the as-deposited $W_2B$ at 25 °C and ~0.45 eV at 150 °C. The data can be fit to yield a negative temperature coefficient for barrier height of $8 \times 10^{-4}$ eV/°C over the range 25-150 °C. The forward I-V characteristics in each case showed the ideality factor was > 2, suggesting transport mechanisms other than thermionic emission, such as recombination. The reverse breakdown voltage also shows a negative temperature coefficient, which may be due to contributions from the reduced barrier height and also from the high defect density in the heteroepitaxial GaN on sapphire. Defect-free GaN is expected to exhibit a positive temperature coefficient for breakdown. The reverse leakage depends on both bias and temperature. From a moderately doped sample of the type studied here, we would expect thermionic emission to be the dominant leakage current mechanism. According to image-force barrier height lowering, this leakage current density, $J_L$ can be written as

$$J_L = -J_S \exp\left(\frac{\Delta \phi_B}{kT}\right)$$  \hspace{1cm} (5-2)

where $\Delta \phi_B$ is the image-force barrier height lowering, given by $\left(\frac{eE_M}{4\pi\varepsilon_S}\right)^{1/2}$ where $E_M$ is the electric field strength at the metal/semiconductor interface and $\varepsilon_S$ is the permittivity. The experimental dependence of $J_L$ on bias and temperature is stronger than predicted from Equation
5-2. The large bandgap of GaN makes the intrinsic carrier concentration in a depletion region very small, suggesting that contributions to the reverse leakage from generation in the depletion region are small. Therefore, the additional leakage must originate from other mechanisms such as thermionic field emission or surface leakage.

Figure 5-3 shows the barrier height and reverse breakdown voltage as a function of annealing temperature. For anneals at $\geq 600 ^\circ C$, the rectifying nature of the W$_2$B contacts was significantly degraded due to the formation of $\beta$-phase W$_2$N, as reported previously for W and WSi$_x$ on GaN $^{37,65}$. The improvement in breakdown voltage at intermediate annealing temperatures may be due to annealing of sputter damage in the near-surface of the GaN.

AES depth profiles of the as-deposited and 700 $^\circ C$ annealed contacts are shown in Figure 5-4. The as-deposited layers (Figure 5-4, A) exhibit relatively sharp interfaces, consistent with the excellent surface morphology evident in the SEM picture of Figure 5-1 A. By sharp contrast, the depth resolution associated with the annealed sample (Figure 5-4, B) is poorer than the as-deposited sample. The bubbling of the film evident in the SEM of Figure 5-1 B might be the source of the degradation in depth resolution. The Ti becomes oxidized upon annealing and separate x-ray diffraction experiments showed the formation of $\beta$-phase W$_2$N at this temperature. Since this phase has been associated with improved ohmicity of W-based contacts on GaN $^{37,65}$, it is no surprise that the rectifying nature of the contact is degraded at this temperature.

**ZrB$_2$ Based Rectifying Contacts**

Figure 5-5 shows the I-V characteristics from the ZrB$_2$/Ti/Au/GaN diodes as a function of post-deposition annealing temperature. The as-deposited sample displays an almost symmetrical characteristic, suggesting that sputter damage dominates the current transport. With subsequent annealing even at 200 $^\circ C$, the reverse breakdown voltage is improved and higher temperatures
increase the reverse current. Figure 5-6 shows the extracted barrier height and reverse breakdown voltage as a function of measurement temperature for as-deposited Zr$_2$B/Ti/Au contacts on n-GaN. From the data, $\phi_b$ was obtained as 0.52 eV for the as-deposited Zr$_2$B at 25 °C, with a maximum value of 0.55 eV after annealing at 200 °C. The barrier height stayed above 0.5 eV until at least 700 °C anneal temperature. While higher barrier heights would be desirable for HEMT operation, there may be applications for the ZrB$_2$ where extended high temperature operation is the most important requirement. In Figure 5-6, the improvement in breakdown voltage at intermediate annealing temperatures may be due to annealing of sputter damage in the near-surface of the GaN. The forward I-V characteristics in each case showed the ideality factor was > 2, suggesting transport mechanisms other than thermionic emission, such as recombination. The reverse leakage depended on both bias and annealing temperature. As explained in the case for W$_2$B schottky contact, for the doping level employed here, we would expect thermionic emission to be the dominant leakage current mechanism. Similarly, given the low intrinsic carrier concentration of GaN, the additional leakage must originate from mechanisms such as thermionic field emission or surface leakage.

SEM micrographs of the contact stack morphology are shown in Figure 5-7 as a function of anneal temperature. The contacts retain a smooth morphology even at 500 °C, where the ohmic contacts already shown significant roughening. The corresponding AES surface scans are shown in Figure 5-8. The as-deposited sample shows only Au on its surface, as expected. After the 350 °C anneal, Ti is evident and its concentration is increased after the 500 °C anneal. Table 5-1 summarizes the near-surface composition data obtained from AES. The essential message from this data is that the Ti outdiffuses and becomes oxidized on the surface.
AES depth profiles of the as-deposited 350 °C annealed and 700 °C annealed contacts are shown in Figure 5-9. The as-deposited layers (Figure 5-9, A) exhibit relatively sharp interfaces, consistent with the excellent surface morphology evident in the SEMs. There is a significant amount of oxygen in the ZrB$_2$ layer, consistent with past reports that the borides are getters for residual water vapor in the ambient during sputter deposition$^{47}$. After the 700 °C anneal, the extent of Ti outdiffusion is increased and the Ti becomes oxidized upon annealing. X-ray diffraction experiments with both a powder system and glancing angle, Ω=1°, crystal system did not show any reaction of the ZrB$_2$ with the GaN even at 800 °C, as shown in Figures 5-10 and 5-11. This is in sharp contrast to the case of W$_2$B contacts on GaN, where β-phase W$_2$N is formed for anneals at ≥ 600 °C. In this latter case, the rectifying nature of the W$_2$B contacts was significantly degraded due to the formation of β-phase W$_2$N, as reported previously for W and WSi$_x$ on GaN$^{37,65}$. Since this phase has been associated with improved ohmicity of W-based contacts on GaN$^{37,65}$, the degradation of rectifying behavior of the contact is expected at this temperature. However, the ZrB$_2$ shows a much slower reaction with the GaN than W$_2$B and the XRD spectra show no formation of nitride or gallide phases.

**Summary and Conclusions**

In conclusion, ZrB$_2$ and W$_2$B exhibits a barrier height of ~0.5 eV on GaN. This is rather low for HEMT gates, but it may have use in applications where thermal stability is more important than gate leakage current such as HEMT gas sensors. The ZrB$_2$/GaN interface is stable against annealing at 800 °C and the contact stability is still determined by outdiffusion of Ti from the ZrB$_2$/Ti/Au stack. Both borides appear to be an efficient getter of water vapor during sputter deposition.
Table 5-1. Near-surface composition data obtained from AES measurements.

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<tr>
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<th>Ti(2)</th>
<th>Au(3)</th>
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<td>[0.212]</td>
<td>[0.000]</td>
<td>[0.049]</td>
</tr>
<tr>
<td>As deposited</td>
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<td>nd</td>
<td>49</td>
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<td>Annealed @350 °C</td>
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<td>16</td>
<td>3</td>
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<td>Annealed @700 °C</td>
<td>45</td>
<td>31</td>
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Figure 5-1. SEM micrographs of W$_2$B based schottky contacts. A) As-deposited. B) Annealed at 700 °C. The inner circle is the W$_2$B/Ti/Au while the outer ring is the ohmic contact.
Figure 5-2. Barrier height and reverse breakdown voltage as a function of measurement temperature for as-deposited W$_2$B/Ti/Au contacts on n-GaN.
Figure 5-3. Barrier height and reverse breakdown voltage as a function of annealing temperature for W$_2$B/Ti/Au contacts on n-GaN.
Figure 5-4. AES depth profiles of W₂B/Ti/Au on GaN. A) Unannealed. B) After annealing at 700 °C.
Figure 5-5. I-V characteristics from ZrB$_2$/GaN diodes as a function of post-deposition annealing temperature.
Figure 5-6. Barrier height and reverse breakdown voltage as a function of annealing temperature for ZrB$_2$/Ti/Au contacts on n-GaN.
Figure 5-7. SEM micrographs of ZrB$_2$ based schottky contacts. A) As-deposited. B) Annealed at 350 °C. C) Annealed at 700 °C. The inner circle is the ZrB$_2$/Ti/Au while the outer ring is the ohmic contact.
Figure 5-8. AES surface scans of ZrB$_2$/Ti/Au on GaN. A) As-deposited. B) Annealed at 350 °C. C) Annealed at 700 °C.
Figure 5-9. AES depth profiles of ZrB₂/Ti/Au on GaN. A) As-deposited. B) Annealed at 350 °C. C) Annealed at 700 °C.
Figure 5-10. Powder XRD spectrum from ZrB$_2$ on GaN. A) Unannealed. B) After annealing at 800 °C.
Figure 5-11. Glancing angle XRD spectra from ZrB$_2$ on GaN. A) Unannealed. B) After annealing at 800 °C.
CHAPTER 6
ANNEALING TEMPERATURE DEPENDENCE OF TiB$_2$ W$_3$B$_5$ AND CrB$_2$ SCHOTTKY BARRIER CONTACTS ON N-GaN

Introduction

The availability of reproducible schottky contacts on GaN or AlGaN is critical to the operation of AlGaN/GaN high electron mobility transistors (HEMTs) for advanced microwave power amplifiers in radar and communication systems. One of the major remaining hurdles in commercializing reliable HEMT-based systems is the need for thermally stable rectifying contacts, since it is anticipated that some power amplifiers may require operating temperatures up to 300 °C. While it might be expected that metal-oxide-semiconductor (MOS) gates would provide superior thermal stability compared to simple schottky metal gates, this would degrade the rf performance due to the extra capacitance and MOS technology is still at a relatively primitive stage for GaN devices. Some standard metallization systems such as Au on AlGaN show an environmental aging effect. Typical schottky metallization for AlGaN/GaN HEMTs are based on Pt/Au or Ni/Au, with the Au included to reduce the sheet resistance of the contact and to prevent oxidation of the other metal. There is still a need to investigate a wider range of thermally stable schottky contacts on GaN in the search for alternatives to Pt/Au or Ni/Au. Other potentially more thermally stable metallization schemes have been reported based on W or WSi$_x$. These exhibit low barrier heights around 0.4-0.5 eV. Another potential class of thermally stable contacts are those based on borides, which have not attracted much attention for use on GaN. The stoichiometric diborides are thermally stable with very high melting temperatures, well in excess of those of both Ni and Pt. They also exhibit good corrosion resistance but are susceptible to oxidation during thermal processing. This may be countered by depositing an overlayer of a metal such as Au in the same deposition chamber.
In this chapter, the electrical characteristics, annealing and measurement temperature dependence of barrier height and stability of TiB$_2$/Ti/Au, W$_2$B$_5$/Ti/Au and CrB$_2$/Ti/Au contacts on n-GaN were studied. The TiB$_2$ contacts show a maximum barrier height of 0.68 eV after annealing at 350 °C. W$_2$B$_5$ and CrB$_2$ are high temperature stable refractory materials which have not been explored as a contact to GaN. These contacts show a maximum barrier height of 0.65 eV and 0.63 eV, respectively, with a small negative temperature coefficient, and are reasonably stable against annealing to ~350 °C. This barrier height is lower than for Ni or Pt and thus one would need to balance the need for improved thermal stability with the poorer rectifying properties.

**Experimental**

The samples employed were 3 µm thick Si-doped GaN grown by Metal Organic Chemical Vapor Deposition on c-plane Al$_2$O$_3$ substrates. The electron concentration obtained from Hall measurements was ~3x10$^{17}$ cm$^{-3}$ for samples used for TiB$_2$ and ~5x10$^{17}$ cm$^{-3}$ for samples used for W$_2$B$_5$ and CrB$_2$ schottky contacts. The surfaces were cleaned by sequential rinsing in acetone, ethanol and 10:1 H$_2$O: HCl prior to insertion in the sputtering chamber. A metallization scheme of X (500 Å) / Ti (200 Å) / Au (800 Å) was used in all experiments where X was either TiB$_2$, W$_2$B$_5$ or CrB$_2$. The Au was added to lower the contact sheet resistance, while the pure Ti is a diffusion barrier. All of the metals or compounds were deposited by Ar plasma-assisted rf sputtering at pressures of 15-40 mTorr and rf (13.56 MHz) powers of 200-250 W. The sputter rates were held constant at 1.4 Å.sec$^{-1}$ for all of the metals or compounds. The contacts were patterned by liftoff of lithographically-defined photoresist and separate samples were annealed at temperatures of 200,350,500 or 700 °C for 1 min in a flowing N$_2$ ambient in a RTA furnace. For ohmic contacts, we used e-beam deposited Ti (200 Å)/Al (800 Å)/Pt (400 Å)/Au (1500 Å)
annealed at 850 °C for 30 secs prior to deposition of the schottky metallization. A ring contact geometry for the diodes was employed, with the schottky contacts of diameter 50-80 μm surrounded by the ohmic contacts with diameter 250-300 μm. The ohmic-schottky spacing was 10 μm.

Auger Electron Spectroscopy (AES) depth profiling of the as-deposited contacts showed sharp interfaces between the various metals. For the AES analysis, the samples were mounted on a stainless steel puck and placed in the system load-lock. After chamber pump-down, the sample puck was inserted into the analytical chamber and placed in front of the analyzer. The AES system was a Physical Electronics 660 Scanning Auger Microprobe. The electron beam conditions were 10 keV, 1 μA beam current at 30° from sample normal. Charge correction was performed by using the known position of the C-(C, H) line in the C 1s spectra at 284.8 eV. The AES spectrometer was calibrated using a polycrystalline Au foil. The Au $f_{7/2}$ peak position was determined to be 84.00±0.02. For depth profiling, the ion beam conditions were 3 keV Ar$^+$, 2.0μA (3 mm)$^2$ raster. The quantification of the elements was accomplished by using the elemental sensitivity factors. We also used Scanning Electron Microscopy (SEM) to examine contact morphology as a function of annealing temperature.

The contact properties were obtained from I-V characteristics of the TiB$_2$/Ti/Au, W$_2$B$_5$/Ti/Au or CrB$_2$/Ti/Au diodes measured over the temperature range 25-150 °C using a probe station and Agilent 4145B parameter analyzer. We fit the forward I-V characteristics to the relation for the thermionic emission over a barrier

$$J_F = A^* T^2 \exp\left(-\frac{e\phi_b}{kT}\right) \exp\left(\frac{eV}{nkT}\right)$$

(6-1)

where $J$ is the current density, $A^*$ is the Richardson’s constant for n-GaN (26.4 A·cm$^{-2}$·K$^{-2}$), $T$ the absolute temperature, $e$ the electronic charge, $\phi_b$ the barrier height, $k$ Boltzmann’s constant.
n the ideality factor and $V$ the applied voltage. The reverse breakdown voltage was defined as the voltage at which the current density was 1 mA.cm$^{-2}$.

**Results and Discussion**

**TiB$_2$ Based Schottky Contact**

I-V characteristics obtained from the diodes annealed at different temperatures are shown in Figure 6-1. The extracted barrier height and reverse breakdown voltage as a function of annealing temperature for TiB$_2$/Ti/Au contacts on n-GaN derived from this data are shown in Figure 6-2. From the data, $\phi_b$ was obtained as 0.65 eV for the as-deposited (i.e. control sample) TiB$_2$ at 25 °C. The barrier height increases with anneal temperature up to 350 °C, reaching a maximum value of 0.68 eV. The work function of sputtered TiB$_2$ is not available, but both Ti and B have work functions around 4.3 eV, while the electron affinity of GaN is 4.1 eV and thus we might expect a low intrinsic barrier height for the compound on GaN. The work function of chemically vapor deposited TiB$_2$ is reported to be in the range 4.75-5 eV $^{47, 85}$. We would also expect that the contact properties of the as-deposited compound would be dominated by residual sputter-damage from the deposition of the contacts (which tends to increase the near-surface n-type conductivity) and once this is annealed out (at ~600 °C in this case), the intrinsic contact properties are revealed. Higher anneal temperatures led to a reduction in barrier height, most likely associated with the onset of metallurgical reactions with the GaN. The reverse breakdown voltage shows a similar trend to the barrier height, going through a maximum where the barrier height is also a maximum. The forward I-V characteristics in each case showed the ideality factor was in the range 2-2.5, suggesting transport mechanisms other than thermionic emission, such as recombination and surface contributions.

Figure 6-3 shows the AES surface scans from these same samples, confirming the onset of Ti outdiffusion by 350 °C. After the 700 °C anneal, the more extensive Ti outdiffusion leads to
oxidation of the top surface of the contact. This is reflected in the summary of the near-surface composition data in Table 6-1. Note that there is also a small amount of Ga outdiffusion from the GaN to the surface. The carbon signal comes from adventitious carbon on the surface.

AES depth profiles of the as-deposited and annealed contacts are shown in Figure 6-4. The as-deposited layers (Figure 6-4, A) exhibit relatively sharp interfaces, consistent with the good surface morphology evident in the SEM pictures described later. The depth resolution of the 350 °C annealed sample (Figure 6-4, B) is slightly poorer than the as-deposited sample, with clear outdiffusion of Ti. After 700 °C annealing, the Ti shows more significant outdiffusion to the surface where it becomes oxidized. The change in interface abruptness at the metal/GaN interface suggests the change in effective barrier height at higher annealing temperatures may result from reactions at the TiB$_2$/GaN interface. This is consistent with the outdiffusion of Ga seen in the AES surface scans. Note also that the oxygen signal increases on the annealed contacts, consistent with the past observation that boride contacts are susceptible to oxidation\textsuperscript{47, 85}. This occurs even though the annealing environment was purified, filtered N$_2$. It is not clear what effect this has on the contact properties, although samples annealed in Ar environments and left to cool completely before removal from the RTA furnace showed similar electrical contact properties and thus to first order, a small amount of oxidation may not be that critical in changing the contact properties.

Figure 6-5 shows SEM images of the contacts both before (A) and after annealing at either 350 (B) or 700 °C (C). The inner contact is the TiB$_2$/Ti/Au, while the outer ring is the Ti/Al/Pt/Au ohmic contact. To the resolution of the SEM, the morphology does not change over this annealing range, although the oxidation of the Ti after the 350 and 700 °C anneal leads to a darker appearance of the rectifying contact.
The I-V characteristics of the as-deposited contacts were measured as a function of measurement temperature up to 150 °C. The extracted barrier height showed no measurable temperature dependence in the range available to us, as shown in Figure 6-6. The reverse breakdown voltage also shows very little temperature dependence. The reverse leakage was found to depend on both bias and temperature. From a moderately doped sample of the type studied here, we would expect thermionic emission to be the dominant leakage current mechanism. The experimental dependence of reverse current on bias and temperature was stronger than predicted from the TiB$_2$ barrier height. The additional leakage must originate from other mechanisms such as thermionic field emission or surface leakage since the large bandgap of GaN makes the intrinsic carrier concentration in a depletion region very small implying generation currents in the depletion region are small. While the initial results with TiB$_2$ show reasonable thermal stability, there is much more additional work that needs to done to establish the long-term reliability of the contacts for HEMT power amplifier or other device applications. This would include additional studies of the interfacial reactions occurring and the effect of bias or environmental-aging in humid ambient.

**W$_2$B$_5$ Based Schottky Contact**

Figure 6-7 shows SEM image of the contacts both before (A) and after annealing at either 350 (B) or 700 °C (C). The inner contact is the W$_2$B$_5$/Ti/Au, while the outer ring is the Ti/Al/Pt/Au ohmic contact. The morphology does not change tremendously over this annealing range. More detailed information on contact reactions can be obtained from the AES measurements. Table 6-2 shows the surface survey data. It is clear from this data that Ti shows some outdiffusion through the Au at 350 °C and this is more significant after 700 °C anneals.

AES depth profiles of the as-deposited and annealed contacts are shown in Figure 6-8. The as-deposited layers (Figure 6-8 A) exhibit relatively sharp interfaces, consistent with the good
surface morphology evident in the SEM picture of Figure 6-7 A. The depth resolution of the 350 °C annealed sample (Figure 6-8 B) is slightly poorer than the as-deposited sample. After 700 °C annealing, the Ti shows significant outdiffusion to the surface where it becomes oxidized upon annealing. Separate x-ray diffraction experiments showed the formation of β-phase W₂N at this temperature, as occurs with pure W (and also the related phase W₂B). The W₂N phase has been associated with improved ohmicity of W-based contacts on GaN 38, 67.

Figure 6-9 shows the I-V characteristics obtained from the diodes annealed at different temperatures. The extracted barrier height and reverse breakdown voltage as a function of annealing temperature for W₂B₅/Ti/Au contacts on n-GaN derived from this data are shown in Figure 6-10. From the data, \( \phi_b \) was obtained as 0.58 eV for the as-deposited W₂B₅ at 25 °C. The barrier height increases with anneal temperature up to 200 °C, reaching a maximum value of 0.65 eV. Higher anneal temperatures led to a reduction in barrier height, most likely associated with the onset of metallurgical reactions with the GaN. The barrier height for the W₂B₅ is higher than for pure W at these moderate anneal temperatures. The reverse breakdown voltage shows a similar trend to the barrier height, going through a maximum where the barrier height is also a maximum.

The I-V characteristics as a function of measurement temperature for as-deposited contacts are shown in Figure 6-11. The reverse leakage was found to depend on both bias and temperature. From a moderately doped sample of the type studied here, we would expect thermionic emission to be the dominant leakage current mechanism 75. According to image-force barrier height lowering, this leakage current density, \( J_L \) can be written as 74

\[
J_L = -J_s \exp \left( \frac{\Delta \phi_B}{kT} \right)
\]  

(6-2)
where $\Delta \phi_B$ is the image-force barrier height lowering, given by \( \left( \frac{eE_M}{4\pi\varepsilon_S} \right)^{1/2} \) where $E_M$ is the electric field strength at the metal/semiconductor interface and $\varepsilon_S$ is the permittivity. The experimental dependence of $J_L$ on bias and temperature is stronger than predicted from equation (2). The large bandgap of GaN makes the intrinsic carrier concentration in a depletion region very small, suggesting that contributions to the reverse leakage from generation in the depletion region are small. Therefore, the additional leakage must originate from other mechanisms such as thermionic field emission or surface leakage.

The extracted barrier height shows only a slight negative temperature coefficient, almost within the experimental error, as shown in Figure 6-12. The forward I-V characteristics in each case showed the ideality factor was $> 2$, suggesting transport mechanisms other than thermionic emission, such as recombination. The reverse breakdown voltage also shows a negative temperature coefficient up to $\sim 100$ °C. Defect-free GaN is expected to exhibit a positive temperature coefficient for breakdown $75$, but we invariably observe negative temperature coefficients in diodes fabricated with any kind of contact on heteroepitaxial GaN on sapphire with its high defect density. The improvement in breakdown voltage at intermediate annealing temperatures may be due to annealing of sputter damage in the near-surface of the GaN $37,65$.

**CrB$_2$ Based Schottky Contact**

Figure 6-13 shows the I-V characteristics obtained from the diodes annealed at different temperatures. The extracted barrier height and reverse breakdown voltage as a function of annealing temperature for CrB$_2$/Ti/Au contacts on n-GaN derived from this data are shown in Figure 6-14. From the data, $\phi_b$ was obtained as 0.52 eV for the as-deposited CrB$_2$ at 25 °C . The barrier height increases with anneal temperature up to 200 °C, reaching a maximum value of
0.62 eV. Higher anneal temperatures led to a reduction in barrier height, most likely associated with the onset of metallurgical reactions with the GaN. The reverse breakdown voltage shows a similar trend to the barrier height, going through a maximum where the barrier height is also a maximum.

AES depth profiles of the as-deposited and annealed contacts are shown in Figure 6-15. The as-deposited layers (Figure 6-15 A) exhibit relatively sharp interfaces, consistent with the good surface morphology evident in the SEM pictures described later. The depth resolution of the 350 °C annealed sample (Figure 6-15 B) is slightly poorer than the as-deposited sample, with clear outdiffusion of Ti. After 700 °C annealing, the Ti shows more significant outdiffusion to the surface where it becomes oxidized.

Figure 6-16 shows the AES surface scans from these same samples, confirming the onset of Ti outdiffusion by 350 °C. After the 700 °C anneal, the more extensive Ti outdiffusion leads to oxidation of the top surface of the contact. This is reflected in the summary of the near-surface composition data in Table 6-3.

Figure 6-17 shows SEM images of the contacts both before (A) and after annealing at either 350 (B) or 700 °C (C). The inner contact is the CrB₂/Ti/Au, while the outer ring is the Ti/Al/Pt/Au ohmic contact. The morphology does not change tremendously over this annealing range, although the oxidation of the contact after the 700 °C anneal leads to a darker appearance of the rectifying contact.

The I-V characteristics were measured as a function of measurement temperature. The extracted barrier height shows very little temperature dependence, almost within the experimental error, as shown in Figure 6-18. The forward I-V characteristics in each case showed the ideality factor was > 2, suggesting transport mechanisms other than thermionic
emission, such as recombination. The reverse breakdown voltage also shows very little
temperature dependence. Defect-free GaN is expected to exhibit a positive temperature
coefficient for breakdown, but we invariably observe negative temperature coefficients in diodes
fabricated with any kind of contact on heteroepitaxial GaN on sapphire with its high defect
density. The reverse leakage was found to depend on both bias and temperature. As explained for
the case of other two boride schottky, the additional leakage must originate from other
mechanisms such as thermionic field emission or surface leakage. The long-term reliability of
these contacts for the HEMT power amplifier applications needs is tested later in the thesis.

Summary and Conclusions

The main conclusions of our study may be summarized as follows:

- $W_2B_5$ produces an as-deposited (by sputtering) barrier height of $\sim 0.58$ eV on GaN and a
  maximum value of 0.65 eV after annealing at 200 °C.

- $TiB_2$ produces an as-deposited (by sputtering) barrier height of $\sim 0.65$ eV on GaN and a
  maximum value of 0.68 eV after annealing at 350 °C.

- $CrB_5$ produces an as-deposited (by sputtering) barrier height of $\sim 0.52$ eV on GaN and a
  maximum value of 0.62 eV after annealing at 200 °C. This is still lower than for Ni or Pt
  HEMT gates, but it may have use in applications where thermal stability is more important
  than gate leakage current such as HEMT gas sensors.

- The Boride/Ti/Au contacts show some outdiffusion of Ti at 350 °C and much more
  significant reaction after 700 °C anneals.

- The as-deposited contacts show only a minor decrease in barrier height for measurement
  temperatures up to 150 °C.

- Additional experiments need to be done to establish the long-term reliability of the contacts
  for the HEMT power amplifier applications.

- The contacts are quite susceptible to oxidation during thermal processing and care must be
  used to minimize exposure to oxidizing ambient.
Table 6-1. Concentration of elements detected on the as-received surfaces of TiB$_2$ based schottky contacts (in Atom %†)

<table>
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<th>Sample ID</th>
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<td>Sensitivity factors</td>
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<td>23</td>
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Table 6-2. Concentration of elements detected on the as-received surfaces of W$_2$B$_5$ based schottky contacts (in Atom %†)

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Table 6-3. Concentration of elements detected on the as-received surfaces of CrB$_2$ based schottky contacts (in Atom %†)

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<td>Annealed at 700 °C</td>
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<td>32</td>
<td>1</td>
<td>17</td>
<td>nd</td>
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</table>

† AES does not detect hydrogen and helium and all concentrations are normalized to 100%. nd = element not detected. AES detection limits range from 0.1 – 1.0 atomic percent.
Figure 6-1. I-V characteristics at 25 °C of TiB$_2$/Ti/Au on GaN as a function of post-deposition annealing temperature.
Figure 6-2. Barrier height and reverse breakdown voltage as a function of annealing temperature for TiB$_2$/Ti/Au contacts on n-GaN.
Figure 6-3. AES surface scans of TiB$_2$/Ti/Au on GaN. A) As-deposited. B) Annealed at 350 °C. C) Annealed at 700 °C.
Figure 6-4. AES depth profiles of TiB$_2$/Ti/Au on GaN. A) As-deposited. B) Annealed at 350 °C. C) Annealed at 700 °C.
Figure 6-5. SEM micrographs of TiB$_2$ based schottky contacts A) As-deposited. B) Annealed at 350 °C. C) Annealed at 700 °C. The inner circle is the TiB$_2$/Ti/Au while the outer ring is the ohmic contact.
Figure 6-6. Barrier height and reverse breakdown voltage as a function of measurement temperature for as-deposited TiB$_2$/Ti/Au contacts on n-GaN.
Figure 6-7. SEM micrographs of W$_2$B$_5$ based schottky contacts. A) As-deposited. B) Annealed at 350 °C. C) Annealed at 700 °C. The inner circle is the W$_2$B$_5$/Ti/Au while the outer ring is the ohmic contact.
Figure 6-8. AES depth profiles of W₂B₅/Ti/Au on GaN. A) As-deposited. B) Annealed at 350 °C. C) Annealed at 700 °C.
Figure 6-9. I-V characteristics of W₂B₅/Ti/Au on GaN as a function of post-deposition annealing temperature.
Figure 6-10. Barrier height and reverse breakdown voltage as a function of annealing temperature for W₂B₅/Ti/Au contacts on n-GaN.
Figure 6-11. I-V characteristics of as-deposited W$_2$B$_5$/Ti/Au on GaN as a function of measurement temperature.
Figure 6-12. Barrier height and reverse breakdown voltage as a function of measurement temperature for as-deposited $W_2B_5/Ti/Au$ contacts on n-GaN.
Figure 6-13. I-V characteristics of CrB$_2$/Ti/Au on GaN as a function of post-deposition annealing temperature.
Figure 6-14. Barrier height and reverse breakdown voltage as a function of annealing temperature for CrB$_2$/Ti/Au contacts on n-GaN.
Figure 6-15. AES depth profiles of CrB$_2$/Ti/Au on GaN. A) As-deposited. B) Annealed at 350 °C. C) Annealed at 700 °C.
Figure 6-16. AES surface scans of CrB$_2$/Ti/Au on GaN. A) As-deposited. B) Annealed at 350 °C. C) Annealed at 700 °C.
Figure 6-17. SEM micrographs of CrB$_2$ based schottky contacts. A) As-deposited. B) Annealed at 350 °C. C) Annealed at 700 °C. The inner circle is the CrB$_2$/Ti/Au while the outer ring is the ohmic contact.
Figure 6-18. Barrier height and reverse breakdown voltage as a function of measurement temperature for as-deposited CrB$_2$/Ti/Au contacts on n-GaN.
CHAPTER 7
IMPROVED LONG-TERM THERMAL STABILITY AT 350°C OF TiB₂ BASED OHMIC CONTACTS ON AlGaN/GaN HIGH ELECTRON MOBILITY TRANSISTORS

Introduction

There is significant interest in developing new metallization schemes for AlGaN/GaN High Electron Mobility Transistors (HEMTs) intended for applications in power amplifiers and converters with high efficiency (above 70%) for radars and communications systems, hybrid electric vehicles, power flow control and remote sensing systems. The achievement of high efficiency microwave operation at elevated temperatures is important from the viewpoint of minimizing the weight and the volume of power stages. AlGaN/GaN HEMTs appear well-suited to simultaneously achieving high powers, high frequencies and high efficiencies. The most commonly used metallization scheme for source/drain contacts on these HEMTs is Ti/Al, with over-layers of Pt, Ni or Ti and then a layer of Au to reduce oxidation problems and lower the sheet resistance of the contact stack. These contacts produce low specific contact resistances when annealed in the 750-900 °C range but there are concerns about the long-term stability during high temperature operation, in part because if the metal layers begin to intermix, a low melting temperature AlAu phase may form that can lead to contact shorting at small electrode separations. One possible solution is to use a very high melting point diffusion barrier in place of the Pt, Ni or Ti in the contact stack. For example, Selvanathan et al. demonstrated that Ti/Al/Mo/Au contacts on n-GaN are stable at 500 and 600 °C for 25 hours of aging, but degraded after 10 hours at 750 °C. We have shown recently that TiB₂, with a melting temperature around 3000 °C and reasonable electrical resistivity (28 μΩ.cm) shows promise as such a diffusion barrier.

In this chapter the long-term aging characteristics at 350 °C of AlGaN/GaN HEMTs with different combinations of TiB₂-based contacts, i.e. those with Ti/Al/TiB₂/Ti/Au source/drain...
metal and Pt/Au or Ni/Au gates and also those with TiB$_2$ diffusion barriers in both the source/drain and gate contacts is studied. Compared with HEMTs with standard Ti/Al/Pt/Au Ohmic contacts and Pt/Au gate contacts, a number of the different combinations of boride-based contacts exhibit superior stability as judged by the change in source-drain current at zero gate voltage and the transconductance.

**Experimental**

The layer structures were grown on sapphire substrates by Molecular Beam Epitaxy and employed a low temperature AlN (300 Å thick) buffer, 2µm of undoped GaN grown at 750 °C under Ga-rich conditions, 250 Å of undoped Al$_{0.2}$Ga$_{0.8}$N and a 30 Å undoped GaN cap. A growth rate of 0.5-1.0 µm·hr$^{-1}$ was used for all depositions. Mesa isolation was achieved with Cl$_2$/Ar inductively coupled plasma etching (300 W source power, 40 W rf chuck power). Ohmic contacts were formed by lift-off of e-beam deposited Ti/Al/Pt/Au subsequently annealed at 900 °C for 1 min in a flowing N$_2$ atmosphere in an RTA furnace. A metallization scheme of Ti (200 Å)/Al (1000 Å)/ TiB$_2$ (500 Å) / Ti (200 Å) /Au (800 Å) was used for comparison in these experiments. All of the metals were deposited by Ar plasma-assisted rf sputtering at pressures of 15-40 mTorr and rf (13.56 MHz) powers of 200-250 W. The contacts were patterned by liftoff and also annealed at 900 °C for 1 min. The specific contact resistance derived from separate Transmission Line Method measurements was ~2x10$^{-6}$Ω·cm$^2$. Schottky gates (1.5x200µm$^2$) of sputter-deposited Ni/Au, Ni/TiB$_2$/Au, Pt/Au or Pt/TiB$_2$/Au were also patterned by lift-off. The HEMT layout is shown in the schematic of Figure 7-1. The HEMT dc characteristics were measured in dc mode using a HP 4145B parameter analyzer. The rf performance of the HEMTs was characterized with a HP 8723C network analyzer using cascaded probes.
The HEMTs were aged for a period of 25 days at 350 °C on a heater plate in air, with the electrical characteristics measured every 2-3 days. The samples were removed from the heater block, allowed to cool to room temperature and measured before being returned for further aging on the heater. The temperature was measured with a thermocouple attached to the heater.

Micro-Raman scattering measurements to determine stress state in the boride-based ohmic contacts were performed in a backscattering geometry with the 488 nm line of an Ar-ion laser. The laser spot size was ~0.8 μm and the laser power at the sample was ~ 9 mW. The bandgap of GaN is larger than the incident photon energy, which minimizes laser-induced heating. Because of its higher relative intensity in this scattering geometry and its sensitivity to stress, we selected the $E_{2}^{2}$ phonon as a probe to monitor the film stress.

Results and Discussion

Since the contact metal reflected the laser light, it was difficult to get enough signal by Raman Scattering to get a quantitative measure of the residual stress in the contact structure. Figure 7-2 A shows an optical image labeled with the four positions from which Raman spectra were acquired (Figure 7-2 B). At the edge of metal contact, we found that there was tensile stress between the Ti/Al/TiB$_2$/Au and the GaN, but it was of a similar magnitude to the conventional Ti/Al/Pt/Au metallization. No peeling or other manifestations of large residual stress were observed in any of our experiments.

Figure 7-3 shows drain-source current as a function of drain-source voltage ($I_{DS}$-$V_{DS}$) characteristics from an AlGaN/GaN HEMT with conventional Ti/Al/Pt/Au source/drain contacts and Pt/Au gate metal before and after aging for 25 days at 350 °C. These devices showed a significant decrease (~30%) in source-drain current within even 2 days of aging and a similar decrease (~35%) in transconductance. The decrease in current in some of the curves at high
voltages in the un-aged sample appears to be due to discharge of traps. The work on Ti/Al/Mo/Au contacts on n-GaN showed that they degraded by in-diffusion of Mo and Au into the semiconductor and also by oxidation of the contacts. Previous work in this regard on ohmic contacts on GaN have also shown that they degrade by intermixing of the contact metallurgy, probably aided in some cases by grain boundary transport.

Figures 7-4 and 7-5 shows similar $I_{DS}-V_{DS}$ characteristics from HEMTs with the Ti/Al/TiB$_2$/Ti/Au ohmic contacts and either Pt/Au or Pt/TiB$_2$/Au gates (Figure 7-4) or Ni/Au or Ni/TiB$_2$/Au gates (Figure 7-5). Several of these combinations show superior thermal stability over the 25 day aging period to the devices with conventional metallization. The boride-based Ohmic contacts retained a smoother morphology than the Ti/Al/Pt/Au both before and after aging, as measured by both atomic force microscopy and optical microscopy. An example is shown in the optical microscope images of Figure 7-6.

Figure 7-7 summarizes the data for percentage change in saturated drain-source current as a function of aging time for all of the HEMTs with different combinations of contacts. All of the devices with boride-based contacts show superior aging characteristics compared to the conventional devices, with the exception of the HEMTs with Pt/TiB$_2$/Au gate contacts. The devices with Pt/Au gate metal and Ti/Al/TiB$_2$/Ti/Au source/drain contacts show only a 10% decrease in $I_{DS}$ after 25 days at 350 °C. Note that all of the devices appear to show a decrease in current that saturates at different levels depending on the metal. This may indicate that the reaction between the GaN and the gate and source/drain metals is limited by both the thermodynamics of each system and by the thickness of the reacting metal.

Figure 7-8 shows rf data for the devices with conventional Pt/Au gate contacts and Ti/Al/Pt/Au ohmic contacts before aging (A) and for a device with Pt/Au gates and
Ti/Al/TiB$_2$/Ti/Au ohmics after (B) 25 days aging at 350 °C. The cutoff frequency, $f_T$, remains at ~5.5 GHz with the maximum frequency of oscillation, $f_{\text{MAX}}$, at ~30 GHz in both cases. After aging the conventional HEMT, we could not obtain reproducible rf data due to difficulties in making consistent contact to the roughened metal surface.

**Summary and Conclusions**

Preliminary aging data on AlGaN/GaN HEMTs with TiB$_2$ diffusion barriers in the source/drain contacts show a higher resistance to degradation than devices with conventional Ti/Al/Pt/Au contacts. Much more work is needed to determine the contact degradation mechanisms and their activation energies and whether aging under bias makes a difference in the contact reliability. The borides are known to be susceptible to oxidation but the presence of the capping layers may reduce the significance of this issue.
Device Geometry

$L_g = 1.5 \, \mu m$, $L_{gd} + L_{gs} + L_g = 18 \, \mu m$
Width$= 200 \, \mu m$

Figure 7-1. HEMT layout used in these experiments.
Figure 7-2. Study of Raman spectra from Ti/Al/TiB₂/Ti/Au contacts on HEMT wafer. A) Optical micrograph. B) Raman spectra.
Figure 7-3. $I_{DS}$-$V_{DS}$ characteristics from HEMT with conventional Pt/Au gate contacts and Ti/Al/Pt/Au source/drain contacts before and after aging at 350 °C for 25 days.
Figure 7-4. $I_{DS}$-$V_{DS}$ characteristics from HEMT with Ti/Al/TiB$_2$/Ti/Au source/drain contacts. A) Pt/Au gate contacts before and after aging at 350 °C for 25 days. B) Pt/TiB$_2$/Au gate contacts before and after aging at 350 °C for 25 days.
Figure 7-5. $I_{DS}$-$V_{DS}$ characteristics from HEMT with Ti/Al/TiB$_2$/Ti/Au source/drain contacts. A) Ni/Au gate contacts before and after aging at 350 °C for 25 days. B) Ni/TiB$_2$/Au gate contacts before and after aging at 350 °C for 25 days.
Figure 7-6. Optical microscopy images of HEMTs. A) Conventional contacts before aging. B) Boride-based source/drain contacts before aging. C) Conventional contacts after aging at 350 °C for 25 days. D) Boride-based source/drain contacts after aging at 350 °C for 25 days.
Figure 7-7. Percent change in saturated drain/source current from HEMTs with different combinations of contact metal schemes as a function of aging time at 350 °C.
Figure 7-8. RF performance of 1.5 x 200 μm² gate length HEMTs. A) HEMT with conventional metal contacts prior to aging. B) HEMT with Pt/Au gates and Ti/Al/TiB₂/Ti/Au source/drain contacts after aging at 350 °C for 25 days.
CHAPTER 8
Ir BASED SCHOTTKY AND OHMIC CONTACTS ON N-GaN

Introduction

GaN high electron mobility transistor (HEMT) power amplifiers have now entered the commercialization stage for use in wireless communications and military applications. There are also possible applications in improved automotive radar and power electronics for hybrid electric vehicles and in advanced satellite communication systems. GaN HEMTs can provide the high-power, high-efficiency, high-linearity RF power transistors required in base stations for mobile data network services. One of the major issues with some applications for these power amplifiers is the need for very stable ohmic and schottky metal contacts, capable of extended operation at elevated temperatures (typically 200ºC or higher). Most ohmic contact schemes for AlGaN/GaN HEMTs use Ti/Al, with over-layers of Ni, Ti or Pt, followed by Au to reduce sheet resistance and decrease oxidation during annealing to achieve the lowest contact resistance needed to achieve the lowest specific contact resistivity. The formation of TiN\textsubscript{X} phases is integral to the contact formation mechanism. One drawback is the often poor lateral edge definition of the contacts and potential shorting to the gate contact because of flow of the low melting temperature viscous AlAu\textsubscript{4} phase. Similarly, the gate metal must be stable during elevated temperature operation and alternatives to the usual Ni, Pd or Pt with overlayers of Au are attractive. There is continued interest in use of high temperature metals such as W, WSi\textsubscript{X}, W\textsubscript{2}B\textsubscript{5}, Mo, V, Ir, Cu, CrB\textsubscript{2}, ZrB\textsubscript{2} and TiB\textsubscript{2} in both schottky and ohmic contacts in an attempt to improve the long term stability at elevated temperatures.

In this chapter the annealing temperature dependence of contact resistance and contact intermixing of Ti/Al/Ir/Au ohmic and Ir/Au schottky metals on n-type GaN is studied. These contacts show promising long-term stability compared to the existing standard metal contact
schemes for n-GaN. Preliminary studies on Ir-based contacts on HEMTs have shown the potential for improved contact performance $^{42,43}$.

**Experimental**

For Schottky contact studies, the GaN samples consisted of 3 μm thick Si-doped GaN grown by Metal Organic Chemical Vapor Deposition (MOCVD) on c-plane Al$_2$O$_3$ substrates. The electron concentration obtained from Hall measurements was ~5x10$^{17}$ cm$^{-3}$. A metallization scheme of Ir (500 Å)/Au (800 Å) was used in all experiments. The Au was added to lower the contact sheet resistance. For comparison to a more conventional metal scheme, we also fabricated samples with Ni/Au contacts with the same layer thicknesses as the Ir/Au. For ohmic contacts to these samples for Schottky studies, we used the standard e-beam deposited Ti (200 Å)/Al (400 Å)/Pt (200 Å)/Au (800 Å) annealed at 850 °C for 30 secs prior to deposition of the Schottky metallization. A ring-contact geometry for the diodes was employed, with the Schottky contacts surrounded by the ohmic contacts. The inner contact diameter was 75 μm.

For ohmic studies, the samples used were also 3 μm thick Si-doped GaN grown by MOCVD on c-plane Al$_2$O$_3$ substrates. The electron concentration obtained from Hall measurements was ~7x10$^{18}$ cm$^{-3}$. Mesas 1.8 μm deep were formed by Cl$_2$/Ar Inductively Coupled Plasma Etching to provide electrical isolation of the contact pads. All of the metals were deposited by Ar plasma-assisted rf sputtering at pressures of 15-40 mTorr and rf (13.56 MHz) powers of 200-250 W. The contacts were patterned by liftoff and annealed at 500-1000 °C for 1 min in a flowing N$_2$ ambient in a RTA furnace. Conventional Ti (200 Å)/Al (400 Å)/Ni (200 Å)/Au (800 Å) was compared with a scheme in which the Ni was replaced with Ir.

Auger Electron Spectroscopy (AES) depth profiling of the as-deposited contacts showed sharp interfaces between the various metals in all contacts. The AES system was a Physical
Electronics Scanning Auger Microprobe. The electron beam conditions were 10 keV, 1 µA beam current at 30° from sample normal. For depth profiling, the ion beam conditions were 3 keV Ar⁺, 2.0 µA (3 mm)² raster. Prior to AES data acquisition, secondary electron microscopy images (SEMs) were obtained from the sample. The SEMs were obtained at magnifications of 125X, and 1,000X. The SEMs were used to locate and document analysis area locations and to document surface morphology. The quantification of the elements was accomplished by using the elemental sensitivity factors.

The Schottky contact properties were obtained from current-voltage (I-V) characteristics of the Ir/Au and Ni/Au diodes measured over the temperature range 25-150 °C using a probe station and Agilent 4145B parameter analyzer. Fitting is done to the forward I-V characteristics to the relation for the thermionic emission over a barrier

\[
J_F = A^* T^2 \exp\left(-\frac{e\phi_b}{kT}\right) \exp\left(-\frac{eV}{nkT}\right)
\]

where \(J\) is the current density, \(A^*\) is the Richardson’s constant for n-GaN, \(T\) the absolute temperature, \(e\) the electronic charge, \(\phi_b\) the barrier height, \(k\) Boltzmann’s constant, \(n\) the ideality factor and \(V\) the applied voltage. For ohmic studies, the contact properties were obtained from Circular transmission line method (CTLM) measurements on circular rings with spacing 5, 10, 15, 20, 25, 30, 35, and 40 µm. The outer diameter of the circular pads was fixed at 300 µm and inner diameter varied from 220 to 290 µm. The specific contact resistance, \(\rho_c\), was obtained from the circular TLM measurements with the relationships

\[
R_T = \frac{R_S}{2\pi} \left[ \ln\left(\frac{R_1}{R_O}\right) + \frac{L_1}{R_O} \frac{I_O(R_O/L_T)}{I_1(R_1/L_T)} + \frac{L_T}{R_1} \frac{K_O(R_O/L_T)}{K_1(R_1/L_T)} \right]
\]

\[
\rho_c = R_S L_T^2
\]

where \(R_T\) is the total resistance, \(R_S\) is the sheet resistance, \(R_1\) is the outer radius of the annular gap, \(R_O\) is the inner radius of the annular gap, \(I_O, I_1, K_O,\) and \(K_1\) are the modified Bessel
functions, \( L_T \) is the transfer length, and \( \rho_c \) is the specific contact resistance. The sheet resistance can be obtained by iterative mathematical process.

**Results and Discussion**

**Schottky Contacts**

Figure 8-1 shows the I-V characteristics obtained from the Ir/Au diodes annealed at different temperatures. The extracted barrier heights as a function of annealing temperature derived from this data are shown in Figure 8-2. From the data, \( \Phi_b \) was obtained as 0.42 eV for the as-deposited Ir at 25 °C. The barrier height increases with anneal temperature up to 500 °C, reaching a maximum value of 0.55 eV in the range 500-700 °C. The barrier height of Ni/Au contacts was of similar magnitude in this annealing range (0.52 to 0.56 eV). Higher anneal temperatures led to high leakage currents in the Ni/Au contacts, associated with the onset of metallurgical reactions with the GaN. By contrast, the Ir contacts did not show the onset of leakage until anneal temperatures of >900 °C. The forward I-V characteristics showed the ideality factor was always higher for Ni, ranging from 1.75 for anneals below 350 °C to > 2 at higher temperatures, compared to 1.3 for Ir at anneal temperatures below 350 °C and 1.8 between 500-700 °C.

AES depth profiles of the annealed Ir/Au and Ni/Au contacts are shown in Figure 8-3. The as-deposited layers exhibited sharp interfaces in both cases. The depth resolution for the 350 °C annealed samples is similar to that of the as-deposited samples. After 700 °C annealing, the Ni/Au contact shows significant outdiffusion of Ni to the surface, corresponding to roughening of the contact morphology. By sharp contrast, the Ir/Au contact shows very little change after 700 °C annealing.
Ohmic Contacts

Figure 8-4 shows the specific contact resistances as a function of anneal temperature for the two ohmic metal schemes. The contact resistance decreased up to ~ 900 °C in both cases, with a lowest contact resistance of $1.6 \times 10^{-6} \, \Omega \, \text{cm}^2$. The minimum in the contact resistance with annealing temperature is most likely related to the formation of low resistance phases of TiN at the interface with the GaN, as reported for conventional contacts previously $^{25-38}$. Annealing at higher temperatures leads to higher contact resistance, which as will be seen later corresponds to extensive intermixing of the contact metallurgy. The contact properties did not show a significant dependence on annealing time at 900 °C.

Figure 8-5 shows the SEM of the contact morphology for the two metallization schemes after annealing at 500 or 900 °C. The morphology is featureless until 900 °C, which corresponds to the minimum in contact resistance. AES surface scans showed only the presence of carbon, oxygen and gold on the as-deposited surface. The carbon is adventitious and the oxygen originated from a thin native oxide on the Au. Figure 8-6 shows the AES depth profiles for the samples corresponding to the SEM images in the previous figure. The profiles from the samples annealed at 500 °C shows significantly less diffusion of Al through the gold layer to the surface in the case of an Ir interlayer and the Ni itself is mobile at 500 °C. The profiles obtained from the samples annealed at 900 °C shows significant inter-diffusion of all the layers.

Figure 8-7 shows the measurement temperature dependence of the contacts on n-GaN annealed at 850 °C. The contacts showed an increased specific contact resistance in the temperature range 320-500 K, most likely due to increased sheet resistivity of the GaN as the carrier mobility decreases. The doping in the n-GaN is not high enough to have the current flow dominated by tunneling. When the tunneling dominates, the specific contact resistivity ($R_{\text{SCR}}$) is dependent upon doping concentration and is basically independent of temperature, i.e.
where $\phi_B$ is the barrier height, $\varepsilon_S$ the semiconductor permittivity, $m_e^*$ the effective mass of electrons, $\hbar$ the Planck’s constant and $N_D$ is the donor concentration in the semiconductor.

Figure 8-8 shows the room temperature contact resistance of the samples annealed at 900 °C, as a function of aging time spent at 350 °C. This simulates the operation of an uncooled GaN-based transistor and gives some idea of the expected stability of the contact. The conventional Ti/Al/Ni/Au has an increase in specific contact resistance approximately one and a half orders of magnitude after 13.5 days of elevated temperature operation. The contacts showed high resistance (rectifying) behavior beyond this point. By sharp contrast, the Ir-containing contacts show less change with aging time and exhibited a stable specific contact resistance of $\sim 10^{-5} \, \Omega \cdot \text{cm}^2$ after 22 days aging at 350 °C. This suggests that the Ir restricts some of the contact reaction at 350 °C relative to Ni and has a beneficial effect on the long-term stability of the ohmic contacts.

**Summary and Conclusions**

The replacement of Ni by Ir in both ohmic and schottky contacts to n-GaN improves the thermal stability of both types of contacts. The ohmic contacts exhibit superior stability during aging at 350 °C while the schottky contacts show less intermixing of the metals after annealing at 700 °C. The Ir may be a superior choice to boride contact schemes for GaN, since the latter are prone to oxidation.
Figure 8-1. I-V characteristics from Ir/Au Schottky contacts on n-GaN.
Figure 8-2. Schottky barrier height for Ir/Au contacts on n-GaN as a function of annealing temperature.
Figure 8-3. AES depth profiles. A) Ir/Au after annealing at 350 °C. B) Ir/Au after annealing at 700 °C. C) Ni/Au contacts after annealing at 350 °C. D) Ni/Au contacts after annealing at 700 °C.
Figure 8-4. Specific contact resistance of Ti/Al//Ni/Au and Ti/Al/Ir/Au Ohmic contacts on n-GaN as a function of anneal temperature.
Figure 8-5. SEM images Ir and Ni based ohmic. A) Ti/Al/Ir/Au after annealing at 500 °C. B) Ti/Al/Ni/Au after annealing at 500 °C. C) Ti/Al/Ir/Au after annealing at 900 °C. D) Ti/Al/Ni/Au after annealing at 900 °C.
Figure 8-6. AES depth profiles of Ir and Ni based ohmic. A) Ti/Al/Ir/Au after annealing at 500 °C. B) Ti/Al/Ni/Au after annealing at 500 °C. C) Ti/Al/Ir/Au after annealing at 900 °C. D) Ti/Al/Ni/Au after annealing at 900 °C.
Figure 8-7. Specific contact resistance of Ti/Al//Ni/Au and Ti/Al/Ir/Au ohmic contacts on n-GaN as a function of measurement temperature.
Figure 8-8. Specific contact resistance of the Ti/Al/Ni/Au and Ti/Al/Ir/Au contacts annealed at 900°C as a function of aging time at 350 °C.
As discussed earlier, while further improvements in the III-V nitride materials quality can be expected to enhance device operation, further device advances will also require improved processing technology. Owing to their wide bandgap nature and chemical stability, GaN and related materials present a host of device processing challenges. One of the critical areas is thermal processing, high temperature ohmic and schottky contacts which are thermally stable and can at least sustain harsh condition which the device itself is capable of based on its intrinsic properties.

The problem tackled in this work is reliable low resistance, high temperature operational ohmic contact and reliable high temperature stable rectifying contacts. To this end, new material and metallization schemes were explored which would give better ohmic and schottky contacts. In this context, the contacts being better not only mean low ohmic contact resistance or high schottky barrier height, as it used to mean in earlier work but it shall also mean less roughing of contacts, sharp edge acuity, less intermixing of the metallization or even if the intermixing occurs minimal decrease in specific contact resistance. Some contacts fabricated were tested with prolonged heating over a hot plate or in some cases in hot oven.

In first section, ohmic contact formation on n-GaN using novel Titanium/Aluminum/Tungsten Boride / Titanium / Gold and Titanium / Aluminum / Zirconium Boride / Titanium / Gold metallization schemes were studied using contact resistance, scanning electron microscopy and Auger Electron Spectroscopy measurements. For the case of Tungsten Boride based contact, a minimum specific contact resistivity of $7 \times 10^{-6} \ \Omega \text{cm}^2$ was achieved at an annealing temperature of 800 °C. For the case of Zirconium Boride based a minimum specific contact resistivity of $3 \times 10^{-6} \ \Omega \text{cm}^2$ was achieved at an annealing temperature of 700 °C. This order of specific contact
resistance is comparable to that achieved with conventional Ti/Al/Pt/Au on the same samples. The lowest contact resistance was obtained for 60 s anneals. The contact resistance was essentially independent of measurement temperature, indicating that field emission plays a dominant role in the current transport. The Ti began to out diffuse to the surface at temperatures of ~500 °C, while at 800 °C the Al also began to intermix within the contact. By 1000 °C, the contact showed a reacted appearance and AES showed almost complete intermixing of the metallization. The contact resistance showed excellent stability for extended periods at 200 °C, which simulates the type of device operating temperature that might be expected for operation of GaN-based power electronic devices.

Keeping the contact at 200°C for prolonged duration was an important step and must be looked as important step towards achieving the goal for all the future study in this area of contact study.

In another section, three different metal borides (TiB₂, CrB₂ and W₂B₅) were examined for use in Ti/Al/boride/Ti/Au ohmic contacts on n-type GaN and the reliability compared to the more usual Ti/Al/Ni/Au metal scheme. A minimum contact resistance of 1.5x10^{-6} Ω.cm² was achieved for the TiB₂-based scheme at an annealing temperature of 850-900 °C. For W₂B₅ the minimum contact resistance was ~1.5x10^{-5} Ω.cm² at 800 °C while for CrB₂ it was 8x10^{-6} Ω.cm² at 800 °C. Thus, minimum specific contact resistance obtained with TiB₂ was approximately an order of magnitude lower than with CrB₂ and W₂B₅. In all cases, the minimum contact resistance is achieved after annealing in the range 700-900 °C. The contact resistance did not change significantly with changing temperature at which the I-V measurements were done. The TiB₂ and CrB₂ contacts retain smooth morphology even after annealing at 1000 °C. Auger Electron Spectroscopy depth profiling indicated that formation of an interfacial TiNx layer is likely
responsible for the ohmic nature of the contact after annealing. All three boride-based contacts were tested at even harsher condition this time, being placed over a hot plate for period of more than 22 days at temperature of 350 °C. After extended aging the boride based contacts, in general, show less change in specific contact resistance than Ti/Al/Ni/Au even as the intermixing of the metallization scheme occurs.

Then, Schottky contact formation on n-GaN using a novel W₂B based and Zirconium based metallization scheme was studied using current-voltage, scanning electron microscopy and Auger Electron Spectroscopy measurements. A maximum barrier height of ~0.55 eV was achieved on as-deposited samples for Tungsten boride based and after 200 °C anneal for Zirconium Boride based, with a negative temperature coefficient of 8 x 10⁻⁴ eV/°C over the range 25-150 °C. The barrier height was essentially independent of annealing temperature up to 500 °C for W₂B and 700 °C for ZrB₂ and decreased thereafter due to the onset of metallurgical reactions with the GaN. The Ti began to outdiffuse to the surface at temperatures of >500 °C. In conclusion, two borides produces only a low barrier height of ~0.5 eV on n-GaN. This is rather low for HEMT gates, but it may have use in applications where thermal stability is more important than gate leakage current such as HEMT gas sensors.

In the next section, the annealing temperature (25-700 °C) dependence of schottky contact characteristics on n-GaN using TiB₂, CrB₂ and W₂B₅ based metallization scheme deposited by sputtering are reported. The main conclusions of this study may be summarized as follows:

- W₂B₅ produces an as-deposited (by sputtering) barrier height of ~0.58 eV on GaN and a maximum value of 0.65 eV after annealing at 200 °C.
- TiB₂ produces an as-deposited (by sputtering) barrier height of ~0.65 eV on GaN and a maximum value of 0.68 eV after annealing at 350 °C.
- CrB₂ produces an as-deposited (by sputtering) barrier height of ~0.52 eV on GaN and a maximum value of 0.62 eV after annealing at 200 °C. This is still lower than for Ni or Pt.
HEMT gates, but it may have use in applications where thermal stability is more important than gate leakage current such as HEMT gas sensors.

- The Boride/Ti/Au contacts show some outdiffusion of Ti at 350 °C and much more significant reaction after 700 °C anneals.
- The as-deposited contacts show only a minor decrease in barrier height for measurement temperatures up to 150 °C.
- Additional experiments need to done to establish the long-term reliability of the contacts for the HEMT power amplifier applications.
- The contacts are quite susceptible to oxidation during thermal processing and care must be used to minimize exposure to oxidizing ambient.

AlGaN/GaN High Electron Mobility Transistors (HEMTs) were fabricated with Ti/Al/TiB₂/Ti/Au source/drain ohmic contacts and a variety of gate metal schemes (Pt/Au, Ni/Au, Pt/TiB₂/Au or Ni/TiB₂/Au) and subjected to long-term annealing at 350 °C. By comparison with companion devices with conventional Ti/Al/Pt/Au ohmic contacts and Pt/Au gate contacts, the HEMTs with boride-based ohmic metal and either Pt/Au, Ni/Au or Ni/TiB₂/Au gate metal showed superior stability of both source-drain current and transconductance after 25 days aging at 350 °C.

Ir/Au schottky contacts and Ti/Al/Ir/Au ohmic contacts on n-type GaN were investigated as a function of annealing temperature and compared to their more common Ni-based counterparts. The Ir/Au ohmic contacts on n-type GaN with n~ 10¹⁷ cm⁻³ exhibited barrier heights of 0.55 eV after annealing at 700 °C and displayed less intermixing of the contact metals compared to Ni/Au. A minimum specific contact resistance of 1.6 x10⁻⁶ Ω.cm² was obtained for the ohmic contacts on n-type GaN with n~10¹⁸ cm⁻³ after annealing at 900 °C. The measurement temperature dependence of contact resistance was similar for both Ti/Al/Ir/Au and Ti/Al/Ni/Au, suggesting the same transport mechanism was present in both types of contacts. The Ir-based ohmic contacts displayed superior thermal aging characteristics at 350 °C. Auger Electron
Spectroscopy showed that Ir is a superior diffusion barrier at these moderate temperatures than Ni.
LIST OF REFERENCES


BIOGRAPHICAL SKETCH

Rohit Khanna was born on 25th February, 1981 in Lucknow, Uttar Pradesh, India. He grew up and spent his high school years in Varanasi, Uttar Pradesh, India. On graduating from high school in 1999, he secured a position in Indian Institute of Technology-Joint Entrance Examination (IIT-JEE), earning an admission to the prestigious Institute of Technology, Banaras Hindu University (IT-BHU), Varanasi, India.

He obtained his Bachelor of Technology (B.Tech.) from the Department of Ceramic Engineering at IT-BHU in 2003. Then he applied for graduate studies and was accepted in the Department of Materials Science and Engineering (MSE) at UCLA and UFL (USA). In fall 2003 he joined the doctoral program at the MSE at the University of Florida. He joined Prof. Dr. Pearton’s research group from spring 2004. While continuing his doctorate program under Prof. Dr. Pearton, he was offered a job with Oerlikon USA Inc (earlier Unaxis and Plasma Therm) which he joined as an Associate Applications Lab Engineer in January 2007.