Rigorous Investigation of AlGaN/GaN Heterostructure Surface Treatments with Si Thin Films

by

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Abstract

Wide bandgap AlGaIn/GaN High Electron Mobility Transistors (HEMTs) are currently the subject of intensive research and have already demonstrated an unprecedented performance in high power, high frequency and high temperature areas of applications. In order to improve ohmic contacts and to address the performance degradation due to surface states on this material, several research groups have used Si-containing films on the semiconductor surface.

In this study a rigorous investigation of the “treated” and “coated” AlGaIn/GaN surfaces with two different Si-containing thin films was carried out. “Treatment” is defined as the deposition of Si-containing thin film, encapsulation with SiO₂, Rapid Thermal Annealing (RTA) and the films removal with a wet etch. “Coating” is defined as the deposition of Si film on the AlGaIn/GaN surface left in place for all further processing. Treated, coated and control samples were examined with Secondary Ion Mass Spectroscopy (SIMS), X-ray Photo-electron Spectroscopy (XPS), Hall and contact angle measurements in order to establish the physical mechanism responsible for the electrical changes on the Si-treated surfaces. Previous workers have attributed the improvement in the electrical properties to deep diffusion of Si species into the GaN or AlGaIn epilayers, as indicated by SIMS profiling. In this work, it is demonstrated that this detection of Si was due to a SIMS profiling artifact.

A remnant SiN_x film with the approximate thickness between 1.0 and 2.0 nm was identified by SIMS and XPS on the Si-treated AlGaIn surfaces after the Si film was

removed with an aggressive HF:HNO₃ wet etch. The water contact angle measurement has revealed the similarity between Si-coated and Si-treated samples, and the results were in sharp contrast to the measurements on the control samples.

The Si treatments and coatings were found to pre-passivate the AlGaN surface, significantly improving the Two-Dimensional Electron Gas (2DEG) sheet resistances and current densities before silicon nitride passivation. The best ohmic contact resistance values were not found to be affected significantly by the treatments or coatings and remained in the range of 0.5 - 0.6 Ω -mm ($\rho_c \approx 6.3 \times 10^{-6}$ - 9.0×10^{-6} Ω -cm²). However, Si treatments and coatings were found to significantly improve the ohmic contact resistance values in adverse or poorly optimized processing conditions. Both the treatments and the coatings have resulted in higher leakage currents and more negative threshold voltages in fabricated devices.

It was concluded that the solid state diffusion was not responsible for the electrical enhancements in fabricated devices. The electrical changes observed in this study were attributed to the remnant SiN_x film on Si-treated and Si film on Si-coated AlGaN surfaces.

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Nomenclature

2-DEG	Two-Dimensional Electron Gas
AR-XPS	Angular Resolved X-ray Photo-electron Spectroscopy
BOE	Buffered Oxide Etchant
c-TLM	Circular Transmission Line Model
DI	Deionized (Water)
FE	Field Emission
FWHM	Full Width at Half Maximum
IPA	Isopropyl Alcohol
HEMT	High Electron Mobility Transistor
HFET	Heterostructure Field Effect Transistor
HMDS	Hexamethyldisilazane, Adhesion Promoter
ICP	Inductively Coupled Plasma
MBE	Molecular Beam Epitaxy
MOCVD	Metal-Organic Chemical Vapour Deposition
NRC	National Research Council
PECVD	Plasma-Enhanced Chemical Vapour Deposition
RTA	Rapid Thermal Annealing
SBH	Schottky Barrier Height
SIMS	Secondary Ion Mass Spectroscopy
TE	Thermionic Emission
TFE	Thermionic Field Emission
TLM	Transmission Line Model

WZ	Wurtzite, Crystal Structure
XPS	X-ray Photo-electron Spectroscopy
ZB	Zincblende, Crystal Structure

Chapter 1

INTRODUCTION

1.1 Motivation

The wide bandgap GaN semiconductor material has generated an enormous amount of interest in recent years due to its unique properties and unchallenged performance in high power, high temperature and high frequency applications. Due to the wide bandgap, high electron mobility, high breakdown electric field and unusually strong polarization properties, AlGaN/GaN based High Electron Mobility Transistors (HEMTs) have reached power densities of 30-40 W/mm, more than one order of magnitude higher than GaAs. At the same time, unity gain frequency f_T and maximum oscillation frequency f_{max} of about 200 GHz have been achieved, extending the application of GaN devices to millimetre wave and beyond [1].

Making good ohmic contacts to GaN and AlGaN/GaN heterostructures remains a major challenge to this day. Research aimed at improvements to ohmic contacts and/or sheet carrier densities in AlGaN/GaN heterostructures, as well as better insights into the physical mechanisms underlying these enhancements, would aid substantially to the development of superior GaN electronic devices.

One proposed method to improve ohmic contacts to GaN or AlGaN/GaN material relies on the apparent doping of the surface with a thin Si-containing film using high

temperature treatments (referred to as Si “treatment”). Evidence for this supposed solid state diffusion of Si was limited to Secondary Ion Mass Spectroscopy (SIMS) profiles obtained in previous work, which is reviewed in Section 2.2. Appropriate control samples for SIMS measurements were not included or analyzed. Thus, previously reported Si concentration depth profiling results obtained by SIMS were likely to contain a primary ion forward scattering artifact, making the Si diffusion appear deeper. The mechanisms by which the process steps involved in Si treatments could improve the electrical performance of AlGaIn/GaN heterostructures were not fully investigated and properly identified. The necessity of Si diffusion for better ohmic contacts is questionable and was not addressed by the previous research. In addition, carrier densities after Si wet etch and subsequent HEMT fabrication were not previously reported.

1.2 Thesis Objectives

The primary objective of this work was to identify the main mechanism behind the improvements to ohmic contacts and carrier sheet densities in AlGaIn/GaN material achieved by Si treatments. The second objective of this work was to investigate AlGaIn/GaN structures on which the Si film was intentionally left in place for all further processing (referred to as Si-“coated”), and compare them to the Si-treated material.

1.3 Thesis Organization

An overview of GaN materials and devices, as well as a review of previous work on the use of Si on GaN and AlGaIn surfaces are given in Chapter 2. Process flow considerations and design as well as the performed electrical and chemical composition

measurements are given in Chapter 3. Chapter 4 briefly explains and presents each measurement result. Chapter 5 provides discussion of the results. Finally, thesis summary and areas for future work are given in Chapter 6.

1.4 Thesis Contributions

As a result of this research, it was established that deep diffusion of Si into AlGaIn layer is not responsible for the enhancement to the electrical performance of AlGaIn/GaN devices. A remnant SiN_x film with the approximate thickness between 1.0 and 2.0 nm, produced by the solid state reaction, remained on AlGaIn surface even after the aggressive HF:HNO_3 wet etch of Si and SiO_2 . The film was analyzed by X-ray Photo-electron Spectroscopy (XPS), SIMS and water contact angle measurements. The forward scattering of Si by the primary ion beam artifact during SIMS profiling was demonstrated. The remnant SiN_x and an intentionally left thin film of Si on AlGaIn surface were shown to pre-passivate the AlGaIn material. The treatment was also found to yield better ohmic contact resistance values in adverse or poorly optimized processing conditions.

Chapter 2

GAN TECHNOLOGY OVERVIEW

2.1 GaN Materials and Devices

2.1.1 Material properties and growth

GaN is a direct wide bandgap semiconductor material with bandgap energy of 3.39 eV at room temperature. The compound is very hard and chemically stable, with the melting point exceeding 2,500°C. However, GaN surface undergoes decomposition at temperatures of about 1,000°C at a pressure of one atmosphere due to nitrogen out diffusion [2].

The wide bandgap allows the material to stay extrinsic for a wider range of temperatures, thus making GaN an ideal candidate for high temperature operation. Operating temperatures of several hundred degrees Celsius have been previously reported for GaN devices and circuits [3]. The critical breakdown electric field in semiconductors scales roughly with the square of the bandgap energy and is estimated to be in excess of 4 MV/cm for GaN, compared with 0.2 MV/cm and 0.4 MV/cm for Si and GaAs respectively, making the material very suitable for high voltage operation [4]. Additionally, an adequate electron mobility and a superior saturation velocity make GaN very suitable for high frequency operation (Fig. 2.1).

Figure 2.1: Electron drift velocity at 300 K in GaN, SiC, Si and GaAs computed using the Monte Carlo technique [4].

Table 2.1 compares measured bulk GaN properties with other semiconductors. GaN properties do not stand out significantly in comparison to other wide bandgap semiconductors such as SiC or diamond. Additionally, hole mobility in the material is exceptionally low, about $10 \text{ cm}^2/(\text{Vs})$ [2], although the material allows for p-n junctions [4]. The main advantage of GaN comes from the material's support for ternary and quaternary compounds and HEMT structures [5]:

Material	GaN	Si	GaAs	3H-SiC	4H-SiC	Diamond
Bandgap (eV)	3.39	1.12	1.42	2.2	3.2	5.45
Bandgap Type	Direct	Indirect	Direct	Indirect	Indirect	Indirect
Electron Mobility ($\text{cm}^2/(\text{Vs})$)	1,300	1,350	8,500	900	800	1,900
Saturation Velocity (10^7 cm/s)	2.5-3	1	1	2	2	2.7
Critical Breakdown Field (MV/cm)	4	0.25-0.3	0.4	1.2	3	5.6
Thermal Conductivity ($\text{W/cm} \cdot \text{K}$)	1.3	1.5	0.43	4.9	4.9	20
Dielectric Constant (ϵ_r)	9.5	11.8	13.9	9.6	9.7	5.5
HEMT structures	yes	no	yes	no	no	no

Table 2.1: Comparison table of several most common semiconductor bulk properties [1].

GaN can form alloys with AlN and InN to form ternary and quaternary compounds which span a continuous range of direct bandgap energies from 0.8 eV for InN to 6.2 eV

for AlN (Fig. 2.2). This allows III-N material system to cover a very broad range of energies and thus emission wavelengths from the infra-red to the deep ultraviolet are unchallenged by any other available material today [4]. The large range of available bandgaps and support for heterostructures in III-V materials also allows for bandgap engineering.

Figure 2.2: Bandgap energy at 300 K as a function of lattice constant of III-N and other semiconductors [5].

GaN, AlN and InN form a Wurtzite (WZ) crystal structure at room temperature (also referred to as α -phase). Each element forms a tetrahedral covalent bond with its neighbouring atoms. The Zincblende (ZB) structure can also be found for GaN and InN in thin films. AlN is metastable and has an indirect bandgap in this crystal form. Its growth is also more complicated [5,6].

WZ crystal structure is a member of hexagonal crystal system (Fig 2.3(a)). The crystal lattice structure of GaN can be described by the length of the hexagonal edge a , the height c of the prism, and the microscopic dimensionless parameter u , which is defined as the length of the bond parallel to the c axis in units of c . These parameters for AlN, GaN, and InN can be found in Ambacher *et al.* [2].

Four Bravais-Miller indices are used to describe directions and planes in hexagonal crystal lattices. WZ structure has polar axes (lack of inversion symmetry). In

particular, the bonds in the $\langle 0001 \rangle$ direction for wurtzite are all faced by nitrogen in the same direction and by the cation in the opposite direction. Both bulk and surface properties can depend significantly on whether the surface is faced by nitrogen or metal atoms. The most common growth direction of hexagonal GaN is normal to the $\{0001\}$ basal plane, where the atoms are arranged in bilayers consisting of two closely spaced hexagonal layers, one with cations and the other with anions, so that the bilayers have polar faces. Thus, in the case of GaN a basal surface should be either Ga- or N-faced. By Ga-faced it is meant that Ga is on the top position of the $\{0001\}$ bilayer, corresponding to $[0001]$ polarity (Fig. 2.3(b)). Ga-faced does not mean Ga-terminated; termination should only be used to describe a surface property. A Ga-face surface might be N-terminated if it is covered with nitrogen atoms, but without flipping the crystal it will never be N-faced. It is, however, important to note that the (0001) and $(000\bar{1})$ surfaces of GaN are inequivalent (by convention, the $[0001]$ direction is given by a vector pointing from a Ga atom to a nearest-neighbour N atom) [2].

Figure 2.3: Hexagonal crystal system and different polarities (Ga- and N-faced) of wurtzite GaN.

Growth of GaN single crystals of useful size remains a major challenge to this day. GaN is therefore usually grown epitaxially as a thin film on top of a range of suitable

substrates. The most common GaN epitaxial growth techniques are Metal-Organic Vapour Phase Epitaxy (MOVPE, also known as MOCVD), Molecular Beam Epitaxy (MBE) and Hydride Vapour Phase Epitaxy (HVPE). The most common substrate materials for GaN are sapphire, 6H-SiC and Si. Some less common materials in use are diamond and AlN. 6H-SiC has a lattice mismatch of only 3.5% with GaN (3.0806 Å versus 3.189 Å for GaN), similar thermal expansion coefficients ($4.46 \times 10^{-6} \text{ K}^{-1}$ versus $5.59 \times 10^{-6} \text{ K}^{-1}$ for GaN) and a very high thermal conductivity coefficient. These properties make 6H-SiC the best substrate for fabrication of high power GaN devices. The concentration of dislocations in even the best GaN epilayers is found to be large, $\sim 10^{10} \text{ cm}^{-2}$. The negative impact of these defects on the operation of GaN transistors appears to be rather limited. As grown GaN is always n-type with carrier concentration $\sim 10^{16} \text{ cm}^{-3}$. Unintentional doping with Si, O or nitrogen vacancies is believed to be responsible for this phenomena [2, 8].

2.1.2 Doping of GaN materials

The numbers of free carriers in a semiconductor is dependent on the ionization energies of the donors/acceptors [9]:

$$n \propto N_d T^{3/2} \exp\left(\frac{-E_i}{kT}\right) \quad (2.1)$$

where k is Boltzmann constant ($1.3806503 \times 10^{-23} \text{ J/K}$), T is the absolute temperature in Kelvins, E_i is the ionization energy, and N_d is the donor/acceptor density.

The two most important elements used for intentional doping of GaN materials are Si and Mg. Si is found to substitute on Al and Ga sites in AlN and GaN materials and acts as a shallow donor with ionization energies of 12-17 meV. The electrically active concentration of the donors/acceptors was found to be half the total dose. Mg acts as a shallow acceptor with ionization energy of 155-165 meV. Hydrogen plays

an important role in passivating the Mg acceptors and Rapid Thermal Annealing (RTA) or electron irradiation is required to activate these dopants after the exposure to hydrogen. Both Si and Mg have moderate to low solubility in GaN and AlN materials [2].

Although donor and acceptor atoms can be incorporated during growth of GaN epilayers, selective area doping is not possible with this method. Additionally, heavy doping during growth increases epilayer defect densities. Ion implantation is an enabling technology for creating selective-area doping. The activation energy E_a is the energy needed for implanted dopants to effectively substitute atomic sites in the crystal. The number of incorporated donors N_d is related to anneal temperature by the following relationship [9]:

$$N_d \propto N_{imp} \exp\left(\frac{-E_a}{kT}\right) \quad (2.2)$$

where N_{imp} is the implanted dopant concentration. Studies showed that good activation efficiency can be obtained at 1,100°C, however ion induced damage cannot be completely removed until higher than 1,300°C temperatures are reached. Self compensation starts to occur at these high temperatures due to Si site switching with nitrogen atoms. Unprotected GaN and AlGaN surfaces undergo a significant degradation at these elevated temperatures. Si activation energies from 3.4 to 5.2 eV were reported [4, 9].

An older, alternative method of semiconductor doping is solid or gas-phase doping, accomplished by establishing a high dopant concentration at the boundary of the semiconductor followed by a thermally driven diffusion. One major advantage of this method is the lack of crystal damage which occurs after ion implantation [10].

From a macroscopic point of view, diffusion can be described by the flow of atoms

from areas of high concentration to areas with low concentration:

$$F = -D \frac{\partial C}{\partial x} \quad (2.3)$$

where F is the flux of atoms in $\text{cm}^{-2} \text{s}^{-1}$, D is the diffusivity in $\text{cm}^2 \text{s}^{-1}$ and $\frac{\partial C}{\partial x}$ is the concentration gradient. From a microscopic point of view, the diffusivity D depends exponentially on temperature T and the diffusion activation energy E_{da} [10]:

$$D = D_o \exp \left(\frac{-E_{da}}{kT} \right) \quad (2.4)$$

where D_o is the diffusion coefficient. From this equation, smaller activation energies and larger diffusion temperatures result in faster diffusions. Several mechanisms can contribute to dopant diffusion at the same time, such as interstitial and substitutional diffusions, in which case Eq. 2.4 acquires additional terms. E_{da} in the case of substitutional diffusion and E_a in the case of ion implanted dopant activation are not the same since the former one is the energy required for continuous substitutional process, and the latter one is the energy required for both an implanted atom (most likely in interstitial space) to acquire a site in the crystal lattice and for the repair of local damage. However, these two energies are related to the bond strength in the material.

Diffusion profiles for two cases resulting in the presence of an unlimited source of dopants (e.g. from a highly doped film) or fixed concentration at the surface (diffusion from a gas ambient with a concentration above the solid solubility of the dopant in the solid) are summarized below [10]:

$$C(x, t) = \frac{C_o}{2} \left[1 - \operatorname{erf} \left(\frac{x}{2\sqrt{Dt}} \right) \right] \quad \text{Infinite source of dopants} \quad (2.5)$$

$$C(x, t) = C_s \left[1 - \operatorname{erf} \left(\frac{x}{2\sqrt{Dt}} \right) \right] \quad \text{Constant surface concentration} \quad (2.6)$$

where $\text{erf}()$ is the error function defined by:

$$\text{erf}(z) = \frac{2}{\sqrt{\pi}} \int_0^z \exp(-x^2) dx \quad (2.7)$$

where C_0 is the concentration of an unlimited dopant source, C_s is the constant surface concentration and t is the diffusion time. $2\sqrt{Dt}$ is termed the diffusion length since it provides an extent of dopant profile motion. Subsequent anneals at different temperature with different anneal times result in $(Dt)_{\text{eff}}$ which is given by [10]:

$$(Dt)_{\text{eff}} = D_1 t_1 + D_2 t_2 + \dots \quad (2.8)$$

Dopants show very little movement in GaN materials even at temperatures of 1400°C, with a diffusivity for Si found to be less than $10^{-13} \text{ cm}^2 \text{ s}^{-1}$ at that temperature [4, 11]. Due to high level of defects present in GaN epitaxial films, a diffusion, proceeding along grain boundaries and dislocations, with lower activation energies and higher diffusion constants is also possible [12, 13].

2.1.3 High electron mobility transistors

The high electron mobility transistors, also known as a Heterostructure Field Effect Transistor (HFET), is a field-effect transistor in which the carriers are confined to a thin sheet which is away from scattering centers. This carrier sheet is often abbreviated as a Two-Dimensional Electron Gas (2DEG) to emphasize the confinement of electrons. This electron confinement is usually achieved near the interface between the epitaxially grown layers of compound semiconductors which have different bandgap energy, so that the interface between them constitutes a heterojunction [14].

The initial concept of the accumulation of charge at a heterojunction interface and its potential for devices was introduced in late 1960s. The development in the

1970s of high precision epitaxial-growth techniques made the fabrication of suitable high-quality heterostructure in the III-V compound semiconductors a reality. In 1978 a first GaAs heterostructure based device was demonstrated at Bell Laboratories [15]. The first AlGaIn/GaN HEMT was demonstrated in 1993 shortly after the first GaN MESFET in 1992 by Asif Khan *et al.* [16,17].

An AlGaIn/GaN HEMT structure can be seen in Fig. 2.4. In this structure SiC

Figure 2.4: AlGaIn/GaN HEMT Structure [18].

provides the substrate for GaN and AlGaIn epilayers. A semi-insulating GaN buffer provides a foundation to grow GaN with relatively low defect densities while at the same time providing electrical isolation between adjacent devices on the chip. Semi-insulating properties are often accomplished by iron doping, which compensates unintentionally incorporated donor impurities such as Si, O as well as native defects. The heterojunction of interest forms between an undoped channel layer of GaN and AlGaIn often with an AlN spacer layer in-between. To form a useful heterojunction, the two materials must either have the same lattice spacing or be able to assume one common spacing. In AlGaIn/GaN case the lattice spacing is not the same and thus strained epitaxial layers result, creating a pseudomorphic structure. The spacer

between the two junctions, AlN, promotes a better quality interface which improves 2DEG confinement and reduces electron scattering in the channel. Alloyed ohmic contacts and a Schottky gate provide access and control over current in the 2DEG. Silicon nitride passivation addresses AlGaIn surface states which negatively impact transistor performance.

To fully understand the role of stress on epilayers in GaN HEMTs and to explain the formation and location of the 2DEG, some fundamental physical properties of III-N materials need to be addressed. The Wurtzite structure is non-centrosymmetric and thus it is compatible with the existence of spontaneous polarization. As-grown III-N material possesses atomic layer-to-layer non-vanishing charge separation even in the absence of an external electric field, which can be denoted macroscopically as a polarization vector $\mathbf{P}$. The value of this spontaneous polarization in III-N materials is very large and increasing from GaN to InN and AlN, due to the increasing non-ideality of the crystal structure ($\mathbf{u}$ increases, $\mathbf{c}/\mathbf{a}$ decreases, and the difference from the ideal ratio of lattice constants of a closed packed hexagonal structure, $\mathbf{c}/\mathbf{a} = 1.6333$, increases as well) [19]. The sign of the spontaneous polarization is found to be negative ($\mathbf{P}$ points into the opposite to $[1\ 0\ 0\ 0]$ direction) in III-Nitride semiconductors. In addition to spontaneous polarization, GaN possesses piezoelectric constants up to ten times larger than in GaAs based crystals. The source of piezoelectric polarization is the change to the lattice constants and to the associated change in $\mathbf{u}$ under the influence of induced strain. The orientation of the piezoelectric and spontaneous polarization is parallel in the case of tensile strain and antiparallel in the case of compressive strain in GaN or AlGaIn layers. The spontaneous, piezoelectric and dielectric constants for AlN, GaN and InN can be found in Ambacher *et al.* [20].

Associated with the gradient of polarization in space is a polarization induced charge density given by $\rho_p = -\nabla \cdot \mathbf{P}$. Fig. 2.5(a) depicts the crystal structure and polarization induced charges in Ga-faced and N-faced AlGaIn/GaN HEMT structures.

Figure 2.5: Polarization and induced charges in crystal structure and epilayers in GaN/AlGaN/GaN Ga- and N-faced HEMTs [20].

At an abrupt interface of a heterostructure, the polarization can decrease or increase within a bilayer, causing a fixed polarization charge density. If the polarization induced charge density is positive ($+\sigma$), free electrons will tend to compensate the polarization induced charge resulting in the formation of a 2DEG assuming that there is a source of electrons. In this case the Fermi level E_F will drop below the triangular quantum well at the AlGaN/GaN interface (Fig. 2.6). For Ga-faced AlGaN/GaN heterostructure, polarization induced charge σ can be expressed as:

$$\sigma = P_{AlGaN}^{SP} + P_{AlGaN}^{PE} - P_{GaN}^{SP} \quad (2.9)$$

Because the bandgap of GaN is smaller than of AlGaN, the 2DEG is always formed on the GaN side next to AlGaN/GaN heterostructure. However, depending on crystal polarity, 2DEGs are expected at the lower AlGaN/GaN and upper GaN/AlGaN interfaces for Ga- and N-face GaN/AlGaN/GaN heterostructures respectively (Fig. 2.5(b)).

The bandgap, spontaneous and piezoelectric polarization constants as well as lattice constants and dopant activation energies depend on Al mole fraction in AlGaN.

Figure 2.6: Conduction band diagram of AlGaN/GaN HEMT [4].

The Al mole fraction x describes the ratio of Al to Ga in AlGaN material: $\text{Al}_x\text{Ga}_{1-x}\text{N}$. Some useful linear and quadratic interpolation models for AlGaN and AlGaN/GaN interface properties are summarized in the following equations¹ [21, 22]:

$$a_{\text{AlGaN}}(x) = [3.189 - 0.086x] \text{ \AA} \quad (2.10)$$

$$c_{\text{AlGaN}}(x) = [5.188 - 0.208x] \text{ \AA} \quad (2.11)$$

$$u_{\text{AlGaN}}(x) = [0.3819x + 0.3772(1 - x) - 0.0032x(1 - x)] \quad (2.12)$$

$$\varepsilon_{\text{AlGaN}}(x) = [0.03x + 10.28] \quad (2.13)$$

$$E_{\text{AlGaN}}^G(x) = [6.13x + 3.42(1 - x) - 1.0x(1 - x)] \text{ eV} \quad (2.14)$$

$$\Delta E^C(x) = [0.63(E^G(x) - E^G(0))] \text{ eV} \quad (2.15)$$

$$\chi_{\text{AlGaN}}(x) = [4.10 - \frac{2}{3}(E^G(x) - E^G(0))] \text{ eV} \quad (2.16)$$

$$P_{\text{AlGaN}}^{SP}(x) = [-0.090x - 0.034(1 - x) + 0.021x(1 - x)] \text{ C/m}^2 \quad (2.17)$$

$$P_{\text{AlGaN}^2/\text{GaN}}^{PE}(x) = [-0.0525x + 0.0282x(1 - x)] \text{ C/m}^2 \quad (2.18)$$

where a and c are lattice constants, u is the dimensionless lattice parameter, ε is the relative permittivity, E^G is the bandgap energy, ΔE^C is the conduction band offset, χ is the electron affinity, P^{SP} and P^{PE} are spontaneous and piezoelectric polarization

¹note there are minor discrepancies for GaN polarization parameters in the literature

²pseudomorphically strained epilayer

constants.

Spontaneous polarization makes a very large contribution to the total polarization in GaN materials. For instance, calculated P^{SP} and P^{PE} values from Eq. 2.17 and Eq. 2.18 for $d_{AlGaN} = 30$ nm and $x = 0.3$ are: $P^{SP} = -1.28 \times 10^{-6}$ C/cm² and $P^{PE} = -0.98 \times 10^{-6}$ C/cm².

The sheet carrier concentration of the 2DEG located at the AlGa_N/Ga_N interface of the nominally undoped structures is given by [23]:

$$n_s(x) = \frac{\sigma(x)}{q} - \frac{\epsilon_0 \epsilon(x)}{dq^2} [q\Phi_B(x) + E_F(x) - \Delta E^C(x)], \quad (2.19)$$

where ϵ_0 is the dielectric constant, d is the AlGa_N barrier width, q is the electron charge, $q\Phi_B(x)$ is the surface barrier height, E_F is the Fermi level position with respect to the Ga_N conduction-band edge, and ΔE^C is the conduction-band discontinuity between AlGa_N and Ga_N. The 2DEG density thus can be directly controlled with Al mole fraction, since both σ and ΔE^C increase with increase in x . The effect of Al concentration on bound interface density and sheet carrier concentration in a Ga-face AlGa_N/Ga_N heterostructure with $d_{AlGaN} = 30$ nm can be seen in Fig. 2.7 [21], in which experimental results are compared with linear and non-linear polarization models. The highest measured and calculated sheet carrier concentration for a pseudomorphic AlGa_N/Ga_N heterostructure is determined to be 2×10^{13} cm⁻² for $x = 0.37$. For higher Al concentration, the AlGa_N barrier starts to relax, diminishing the piezoelectric component.

Some issues still debated for AlGa_N/Ga_N heterostructures include the determination of the Fermi-level position at the free surfaces and the origin of the electrons in the nominally undoped samples accounting for unusually high 2DEG densities. One of the possible models explaining the source of 2DEG is based on the existence of the surface donor-like states, proposed by Mishra [24]. Assuming the donor like states

Figure 2.7: Polarization induced bound interface sheet densities and 2DEG density as a function of Al mole fraction for a Ga-face AlGaIn/GaN heterostructure (30/2,000 nm). Linear and non linear theoretical models and experimental results are compared [21].

are located deep in the AlGaIn bandgap, they will all be occupied at a small value of d (Fig. 2.8(a)). As the thickness of AlGaIn increases, the Fermi level at the surface slides down, approaching and activating deep donors which in turn contribute to the formation of 2DEG. Therefore 2DEG density is expected to depend on the thickness of AlGaIn barrier. This dependence is observed in both metallized and non-metallized structures. Fig. 2.8(b) provides additional theoretical and experimental sheet carrier concentrations dependence on AlGaIn thickness.

For pseudomorphic HEMT structures with barrier thickness of more than 15 nm, the nonlinear dependence of the 2DEG carrier concentration on the alloy composition (Eq. 2.19) can be approximated very well by [21]:

$$n_s(x) = [-0.169 + 2.61x + 4.50x^2]10^{13} \text{ cm}^{-2}, \quad x > 0.06. \quad (2.20)$$

(a) Schematic band diagram illustrating the surface donor model with the undoped AlGaIn barrier thickness less than (a) and greater than (b) the critical thickness for the formation of the 2DEG. Note the position of the Fermi level relative to the surface state in each case. (c) Calculated 2DEG density as a function of barrier thickness according to the surface donor mode [24].

(b) Ambacher *et al.*: Theoretical dependence of the sheet carrier concentration on AlGaIn thickness for $x = 0.15, 0.30$ and 0.45 Al mole fraction. Experimental results for $x = 0.30$ are shown with circles [21].

Figure 2.8: 2DEG density dependence on AlGaIn layer thickness.

The electron mobilities are significantly higher in AlGaIn/GaN heterojunctions than in the bulk GaN material since the 2DEG forms in an undoped GaN material which alleviates the ionized donor scattering mechanism. A thin AlN layer enhances this effect even further by reducing the electron penetration into the AlGaIn due to the higher effective conduction band discontinuity (Fig. 2.9) and also the binary AlN at the interface has no alloy disorder scattering. In addition, the larger conduction band discontinuity results in a small increase in 2DEG concentration [25]. Room temperature mobilities of $2,000 \text{ cm}^2/(\text{Vs})$ are commonly achieved. Very high saturation velocities of $3 \times 10^7 \text{ cm/s}$ are predicted with Monte Carlo simulations (Fig. 2.10).

Figure 2.9: Conduction band diagram of AlGaIn/GaN and AlGaIn/AlN/GaN heterostructures. Dipole in AlN increases the effective ΔE_C (ΔE_C in the figure) [25].

Figure 2.10: Calculated electron velocity vs electric field for AlGaInGaIn HEMT structures with Al mole fractions of 0.15, 0.2 and 0.3 [26].

Very high 2DEG sheet carrier concentration together with the high electron mobilities materialize into a low 2DEG sheet resistance R_s , advantageous for the performance of HEMT devices:

$$R_s = \frac{1}{\mu n_s q} \quad (2.21)$$

Mishra's model provides an insight into the importance of surface states for the 2DEG densities. Regardless of the origin of the source of 2DEG, negative charge

trapped on the surface of AlGaN will negatively affect the 2DEG at the heterostructure by electrostatically repelling it away. Thus, even when the thickness of AlGaN is sufficiently large to support high 2DEG concentrations, the surface conditions can play an enormously important role in the final density of electrons found at the heterojunction.

Figure 2.11: DC and pulsed I-V characteristics of an unpassivated AlGaN/GaN HEMT on SiC substrate. Current collapse (DC to RF dispersion) could be observed in the pulsed mode [1].

Simple analytic charge control models are quite poor for AlGaN/GaN HEMTs, mainly because of many effects present in these devices. The best models rely on numerical solutions to Schroedinger-Poisson equations. AlGaN/GaN HEMTs operation and control equations are similar to that of AlGaAs/GaAs. The dependence of the sheet charge n_s on the applied gate voltage V_g as well as the threshold voltage V_{th} for

AlGaAs/GaAs heterostructures are given by [15]:

$$\begin{aligned} n_s &= \frac{\epsilon}{q(d_d + d_i + \Delta d)}(V_g - V_{th}) \\ &= \frac{\epsilon}{q(d_{barrier} + \Delta d)}(V_g - V_{th}) \end{aligned} \quad (2.22)$$

where

$$V_{th} = \phi_b - \Delta E^C - \frac{qN_d d_d^2}{2\epsilon} + \Delta E_{FO} \quad (2.23)$$

d_d is the AlGaAs donor-layer thickness, d_i is the AlGaAs spacer-layer thickness, ϕ_b is the Schottky barrier height between gate metal and AlGaAs, N_d is the dopant concentration in the donor layer, and Δd and ΔE_{FO} are fitting parameters. Typical AlGaN/GaN HEMTs operate in depletion mode: their threshold (pinch off) voltage is negative. The drain current I_d is directly proportional to the sheet charge density, gate width and drift velocity [27]. Although there are significant physical differences between HEMTs and metal-oxide-semiconductor field-effect transistor, the DC I-V curves are similar for both devices (Fig. 2.11). Pulsed I-V curves are indicative of transistor's RF performance and are prone to a so called "current collapse" phenomena.

2.1.4 Current collapse and silicon nitride passivation

Surface passivation is a fundamental concern in AlGaN/GaN HEMTs technology which alleviates the current-collapse (DC to RF dispersion) phenomena. Current collapse originates from trapping effects at the device surface and/or in the buffer, strongly limiting the device performance under high-frequency operation by reducing the RF transconductance with respect to the DC transconductance [28]. This phenomena can be subdivided into the gate lag and the drain lag related current

collapse.

Fig. 2.11 depicts I-V curves demonstrating the effect of gate lag related current collapse in unpassivated AlGaIn/GaN transistor. Current collapse is mostly eliminated using a thick (~ 100 nm) silicon nitride (Si_3N_4 or SiN) deposited film, which significantly reduces the amount of surface traps available on AlGaIn surface. One of the mechanism thought to be responsible for the reduction of surface traps relies on Si acting as a surface donor and passivating the traps with the donated electrons.

2.1.5 Metal contacts to semiconductor materials

A metal-semiconductor junction can produce two types of contacts: ohmic contacts and rectifying (Schottky) contacts. An ohmic contact displays a linear current/voltage characteristic (Fig. 2.12(a)). A Schottky contact displays a rectifying behaviour, conducting current for only one polarity of voltage bias (Fig. 2.12(b)). Low resistance ohmic contacts are essential for the efficient operation of high power electronic devices, in particular to achieve high current densities, high extrinsic gain and low Joule heating loss to allow high temperature operation [29]. High frequency operation is also degraded by poor ohmic contacts as RC time constant increases.

A Schottky contact with a low reverse bias leakage current and high turn on voltage is essential between the gate metal and the semiconductor for superior HEMT operation. High power diodes have different requirements for the Schottky contact: low reverse bias current, high forward bias current and low turn on voltage.

Depending on the barrier height and width, three electron transport mechanisms can occur at a free metal boundary. If the thermal energy of the electrons at the Fermi level is sufficient to overcome the barrier, Thermionic Emission (TE) occurs. A high barrier sufficiently lowered in the presence of an electric field can give rise to a Thermionic Field Emission (TFE). In the presence of a strong electric field E the potential barrier outside the surface may bend sufficiently steeply and can become

Figure 2.12: I-V characteristic curves for ohmic and Schottky contacts.

very narrow to allow high probabilities for the electrons in the metal to tunnel through. This tunnelling is temperature independent and is termed Field Emission (FE). The current densities of escaping electrons for each case are summarized below [30]:

$$J \approx A^{**}T^2 \exp\left(-\frac{q\Phi_m}{kT}\right) \quad \text{for Thermionic Emission} \quad (2.24)$$

$$J \approx A^{**}T^2 \exp\left[-\frac{(q\Phi_m - \beta_S E^{\frac{1}{2}})}{kT}\right] \quad \text{for Thermionic Field Emission} \quad (2.25)$$

$$J \approx CE^2 \exp\left(-\frac{E_c}{E}\right) \quad \text{for Field Emission} \quad (2.26)$$

where $A^{**} = 2\pi qm^*k^2/h^3$ is the effective Richardson constant, $\beta_S = [e^3/4\pi\epsilon_o]^{\frac{1}{2}}$ is the Schottky coefficient, h is the Planck's constant, m^* is the effective electron's mass, q is the electron charge, k is the Boltzmann constant, T is the absolute temperature and Φ_m is the work function of the metal. C and E_c are temperature independent constants that depend on metal work function and electron's effective mass.

When the metal is brought in contact with the surface of an n-type semiconductor the Fermi levels in both materials align. The difference between the metal work function $q\Phi_m$ and semiconductor electron affinity $q\chi_s$ is defined as the Schottky Barrier Height (SBH), $q\Phi_B$ [31]:

$$\Phi_B = \Phi_m - \chi_s \quad (2.27)$$

SBH can be seen as the energy an electron in the metal must have to cross the barrier

into the semiconductor conduction band. On the other hand the built in potential V_{bi} is the barrier for electrons on semiconductor side (Fig. 2.13) [31].

Figure 2.13: Schottky contact formation [31].

If the work function of the metal is higher than that of an n-type semiconductor, a rectifying contact will form. For lightly doped semiconductors ($N < 1 \times 10^{17} \text{ cm}^{-3}$) the main conduction mechanism is thermionic emission, where the carriers have sufficient thermal energy to surmount the Schottky barrier and pass from a material to the other. In this case, the application of thermionic emission theory at the semiconductor junction interface in the presence of voltage V yields the following current density J :

$$J = A^{**} T^2 e^{\frac{-q\Phi_B}{kT}} \left(e^{\frac{qV}{nkT}} - 1 \right) \quad (2.28)$$

where n is the empirical ideality factor. The ideality factor provides a measure of the quality of a metal-semiconductor junction and is highly process dependent. An ideal Schottky diode will have an ideality factor of 1. Any electron transport mechanism other than thermionic will make the ideality factor larger. It is generally accepted that when $n \ll 1.2$, the thermionic model applies and the plot of $\ln(J)$ vs V can be used to accurately extract the saturation current density and the ideality factor [32]. Image force barrier lowering, electron tunnelling and generation-recombination in depletion region are some of the reasons why Schottky diodes often do not exhibit

a well defined saturation current density under reverse bias conditions [33]. Usually the total current can be expressed as a sum of several current components at the junction [34]:

$$I_{TOT} = I_{TE} + I_{FE} + I_{GR} + I_{LK} \quad (2.29)$$

where I_{TE} is the current due to thermionic emission, I_{FE} is field emission current, I_{GR} is generation-recombination current and I_{LK} is the leakage current.

When the metal work function is lower than the work function of the semiconductor, the contact can be made ohmic by band aligning (Fig. 2.14). This is, however, rarely possible due to the lack of suitably low work function metals (Table 2.2).

Material	Work Function (eV)	Resistivity ($\Omega\text{ cm}$)
Ga	3.96	-
Al	4.28	2.65×10^{-6}
Ti	4.33	4.20×10^{-6}
Ni	5.15	6.84×10^{-6}
Au	5.10	2.35×10^{-6}
Ta	4.25	-
Pd	5.12	1.08×10^{-5}
Pt	5.65	1.01×10^{-5}
Ir	5.62	1.00×10^{-5}
Re	4.72	2.25×10^{-4}
TiN	3.74	-
AlGaN	<4.10	-

Table 2.2: Work functions and resistivity of several metals suitable for Schottky junctions [2, 35].

For many III-V materials SBH is almost independent of metal work function due to the formation of interfacial metal-semiconductor dipoles, the presence of surface states with energies in the forbidden gap and the formation of new metal-semiconductor

Figure 2.14: Ohmic contact formation [31].

alloys at the interface [33]. To account for these effects, SBH can be modelled using the following equation:

$$\Phi_{Bn} = \gamma(\phi_m - \chi_s) + (1 - \gamma)(E^G - \phi_o) \quad (2.30)$$

where

$$\gamma = \frac{\epsilon_i}{\epsilon_i + q\delta D_s}$$

δ and ϵ_i is the thickness of the interface layer and its permittivity, D_s is the density of surface states assumed to be uniformly distributed in the energy within the bandgap up to the level ϕ_o measured from the top of the valance band. The dependence of the SBH on the bandgap value is particularly problematic in wide bandgap semiconductors in which the barrier tends to be high regardless of the value of $\phi_m - \chi_s$, so making a good ohmic contact is difficult. At the same time, because of the inapplicability of the simple Schottky Barrier model (Eq. 2.27), it is also very difficult to achieve a high SBH to make a good Schottky diode.

By far the most common way of achieving an ohmic contact between metal and semiconductor is by creating a heavily doped region at the metal-semiconductor interface. High doping reduces the barrier height via barrier image force lowering

$q\Delta\Phi_B$, but the effect is fairly weak, $\Delta\Phi_B \propto N_d^{\frac{1}{4}} V_{bi}^{\frac{1}{4}}$. The doping, however, also reduces the width of the barrier W , where $W \propto N_d^{-\frac{1}{2}}$. For high doping concentrations ($N_d > 1 \times 10^{19} \text{ cm}^{-3}$) the electrons at the Fermi level can have enough energy to tunnel through the barrier and a tunnelling ohmic contact forms (Fig. 2.15). The carrier transport is ruled by the field emission mechanism.

Figure 2.15: Tunnelling ohmic contact band diagram [31].

An intermediate mechanism of the current transport through the barrier occurs for doping levels in the range of 10^{17} to 10^{19} cm^{-3} , for which electrons at higher than Fermi energy levels can tunnel through the barrier, and is termed the thermionic field emission. The parameter which determines the dominant conduction mechanism is [31]:

$$E_{00} = \frac{h}{4\pi} \left(\frac{N_d}{m^* \epsilon_s} \right)^{\frac{1}{2}} \quad (2.31)$$

where h is the Planck's constant, m^* is the effective electron mass and ϵ_s is the dielectric constant of the semiconductor. Electron transport mechanisms are summarized below:

$$N_d \ll 1 \times 10^{17} \text{ cm}^{-3} \quad \rightarrow \quad kT \gg qE_{00} \quad (\text{TE})$$

$$N_d \gg 1 \times 10^{19} \text{ cm}^{-3} \quad \rightarrow \quad kT \ll qE_{00} \quad (\text{FE})$$

$$N_d \approx 1 \times 10^{17} - 1 \times 10^{19} \text{ cm}^{-3} \quad \rightarrow \quad kT \approx qE_{00} \quad (\text{TFE})$$

A general specific contact resistivity at the junction as a function of current density is given by [31]:

$$\rho_c = \left(\frac{\partial J}{\partial V} \right)_{V=0}^{-1} \quad (2.32)$$

When this expression is applied to the thermionic current density across the metal-semiconductor junction given in Eq. 2.28, the following specific contact resistivity is obtained:

$$\rho_{cj} = \frac{k}{qA^{**}T} \exp\left(\frac{q\Phi_B}{kT}\right) \quad (2.33)$$

When thermionic field emission and field emission are considered, the following relations result:

$$\rho_{cj} \propto \exp \frac{\Phi_B}{E_{00} \coth\left(\frac{E_{00}}{kT}\right)} \quad \text{for thermionic field emission} \quad (2.34)$$

$$\rho_{cj} \propto \exp\left(\frac{\Phi_B}{E_{00}}\right) \propto \exp\left(\frac{\Phi_B}{N_d}\right) \quad \text{for field emission} \quad (2.35)$$

As can be seen from these equations, the specific contact resistivity in tunnelling contacts is strongly dependent on both barrier height and doping concentration.

The total contact resistance between the ohmic metal and the channel in the device can be written as:

$$R_{c.tot} = R_{cj} + R_{met} + R_{csem} \quad (2.36)$$

where

$$R_{cj} = \frac{\rho_{cj}}{A_{\text{eff}}} \quad (2.37)$$

is the resistance associated with the metal-semiconductor junction, A_{eff} is the effective contact area, R_{met} is the resistance associated with the ohmic metal and R_{csem} is the resistance associated with the semiconductor material. An ohmic contact will have a negligible R_{cj} resistance in comparison with the other two.

Theoretical calculations do not predict contact resistances well, thus it is often necessary to extract the parameters from measurements on fabricated structures. This is usually done using Transmission Line Model (TLM) structures (Fig. 2.16). The resistance measured between rectangular ohmic patches follows the equation:

Figure 2.16: TLM test structure.

$$R(d) = 2\frac{R_c}{W} + \frac{R_s}{W}d \quad (2.38)$$

where R_c is the width independent ohmic contact resistance, often conveniently expressed in units of $\Omega\text{-mm}$, W is the width of the rectangular contact, R_s is the sheet resistance between the contacts and d is the contact spacing. The fit of a straight line to the measured data is used to extract both R_c and R_s parameters (Fig. 2.17).

For lateral ohmic contacts an effective conducting contact length, or transfer length, is defined as:

$$L_t = \left(\frac{\rho_c}{R_{sc}} \right)^{\frac{1}{2}} \quad (2.39)$$

Figure 2.17: Resistance measured between adjacent TLM structure contacts versus their spacing.

where ρ_c is the specific ohmic contact resistivity and R_{sc} is the sheet resistance of the semiconductor below the contact. Note that R_{sc} does not have to be equal to R_s . Transfer length can be interpreted as the distance at which the current density at the lateral contact drops by a factor of e^1 . The contact resistance can be found by the following relation [36]:

$$R_c = \frac{\sqrt{\rho_c R_{sc}}}{\tanh(\frac{L}{L_t})} \quad (2.40)$$

For $L \gg L_t$, the effective area of the contact is defined by L_t , thus Eq. 2.40 can be rearranged to get:

$$R_c = \frac{\rho_c}{L_t} \quad (2.41)$$

For $L \ll L_t$, the effective area of the contact is defined by L , thus Eq. 2.40 becomes:

$$R_c = \frac{\rho_c}{L} \quad (2.42)$$

A modified TLM method, referred to as circular-TLM (c-TLM), uses circular ohmic pads surrounded by rings of ohmic metal (Fig. 2.18). The advantage of this method lies in avoiding the need for mesa isolation to define the boundaries of the conduction layer.

Figure 2.18: c-TLM structure.

For a transfer length significantly smaller than the dimension of the contact pad r_0 , c-TLM resistance can be modelled using the following equation [37]:

$$\begin{aligned} R(d) &\approx \frac{R_s}{2\pi} \left[\ln \left(\frac{r_0 + d}{r_0} \right) \right] + \frac{R_{sc}}{2\pi} \left[L_t \left(\frac{1}{r_0 + d} + \frac{1}{r_0} \right) \right] \\ &= \frac{R_s}{2\pi} \left[\ln \left(\frac{r_0 + d}{r_0} \right) \right] + \frac{R_c}{2\pi} \left[\frac{1}{r_0 + d} + \frac{1}{r_0} \right] \end{aligned} \quad (2.43)$$

By fitting this equation to c-TLM measured data, contact resistance R_c and the sheet resistance R_s can be extracted.

2.1.6 Ohmic and Schottky contacts to GaN and AlGaIn/GaN

A Ti/Al based metal stack has traditionally been used to achieve good ohmic contacts on GaN and AlGaIn layers. A specific contact resistivity in the range of 10^{-7} to 10^{-6} $\Omega\text{-cm}^2$ was obtained by several groups [29]. Both Ti and Al have a fairly low work function, Ti has a very good adhesion properties to GaN and upon alloying reduces native oxides formed on the GaN surface. Solid phase reaction between Ti and GaN forming TiN is possible since the heat of formation of TiN and GaN are -265 and -110.9 kJ/mol respectively. Out diffusion of nitrogen creates donor vacancies in GaN and AlGaIn [38]. Only 2 monolayers of TiN formed at the interface are sufficient to generate a 10 nm thick layer of GaN with a doping concentration of 10^{20} cm^{-3} which is suitable for tunnelling contacts [39]. In addition, TiN has lower than 4 eV work function, which would assist tunnelling. The addition of Al allows for lower ohmic

contact resistances than those obtained with Ti layer alone. The exact mechanism by which Al reduces the contact resistance is unclear. There are indications that Al reduces the reactivity of Ti towards GaN or AlGaN layers and this prevents undesired contact features, such as voids, from forming [40]. Luther *et al.* [41] proposed that Al diffuses through the Ti layer, resulting in an ohmic contact when a low work function Al-Ti inter-metallic phase comes into contact with the GaN surface.

Both Ti and Al have a high propensity towards oxidation even at low oxygen concentrations. Additionally, Al tends to ball up during alloying (Al melting point is around 660°C [29]), resulting in a rough surface morphology of the Ti/Al contacts. In order to minimize this problem, a Ti/Au overlayer (melting temperatures of 1,600°C and 1,060°C respectively) is commonly used on top of standard Ti/Al metallization. An additional Ti layer helps to tie up excess Al, preventing it from balling up [39]. In general, annealing of the four layer ohmic metal stack results in the formation of many alloys (TiN, TiAl, AlN and Al₂Au) and the final contact resistance is very sensitive to the individual layer thickness as well as anneal conditions [29]. Typical anneal temperatures for Ti/Al/Ti/Au ohmic metal schemes range from 800 to 900°C. As of 2002 Ti/Al/Ti/Au contact resistance ranged from 0.7 to 2.3 Ohm-mm [42]. More recently ohmic contact resistances to GaN and AlGaN material range from 0.5 to 1 Ohm-mm with some experimental stacks reaching 0.3 Ohm-mm and lower values [29].

It should be noted that a trade-off exists between the ohmic contact resistance and the 2DEG sheet resistance values in AlGaN/GaN heterostructures. Thick AlGaN epilayers with high Al mole fractions are beneficial for high 2DEG densities and thus low 2DEG sheet resistances, but disadvantageous for obtaining good ohmic contacts. Generally, the 2DEG sheet resistance values have to be taken into account when comparing the ohmic contact resistance values from different studies directly.

Metals with very high work functions, such as Ni, Pt, Au, Re and Ir (Table 2.2), are most commonly used for Schottky diodes on GaN. Ni is a very commonly used

metal for a rectifying contact on GaN and AlGaN. It forms a barrier height of $\sim 0.84 - 1.11$ eV for Al mole fraction from 0 to 0.23. Ideality factor varies from 1.12 to 1.37 [32]. In multilayer Schottky contact schemes Ni is often used as a first layer adjacent to GaN due to its superior adhesion properties. In the study by Miura *et al.* [43], Ni/Pt/Au gave better Schottky contacts than either Pt/Au or Ni/Au on $\text{Al}_{0.26}\text{Ga}_{0.74}\text{N}$ and n-GaN epilayers. Superior barrier heights of 1.47 eV (n-GaN) and 0.74 eV (AlGaN) as well as the lowest reverse bias leakage currents were achieved. Ni/Pt/Au Schottky diodes were also compared to Ti/Ni/Pt/Au, Pt/Au and Ti/Pt/Au by Miura *et al.* [44]. The Ni/Pt/Au scheme was found superior in both the barrier height and the reverse bias leakage, with $\phi_B = 0.85$ eV and $J_{\text{leak}} = 10^{-2}$ A/cm². Ti/Pt/Au scheme, on the other hand, displayed a very small variation in both leakage current and barrier height with RTA anneal temperatures, which typically range from 400 to 500°C [43].

Low barrier heights and high non-ideal leakage currents are still a problem in Schottky diodes to GaN and AlGaN materials. Some effort has been made in order to understand the mechanism of gate leakage current in AlGaN/GaN HFETs. One possible mechanism found in the paper by Hasegawa *et al.* [45] relies on a thin surface barrier model and assumes the presence of high density of deep donor defects (at $E^C - 0.37$ eV) near AlGaN surface, causing a narrowing of the Schottky barrier in such a way that electrons can tunnel through this barrier in both forward and reverse direction by means of the thermionic field-emission or the field-emission mechanism, depending on the temperature.

2.1.7 Surface preparation, cleaning and chemical analysis

Surface preparation plays an important role in any semiconductor material device fabrication, including that of GaN. In AlGaN/GaN heterostructures, surface states can directly degrade 2DEG density and ohmic contacts. According to Mishra's model,

any impurities on the surface which act as acceptors or deep traps will directly reduce the 2DEG. The same impurities will counter-dope the semiconductor during the formation of an n-type tunnelling contact, which is a highly undesirable effect. The use of Fe, Be, Zn as well as Carbon (C) as acceptor-like point defects was previously demonstrated [46]. Thus particular attention has to be given to organic contamination. In addition, GaN surfaces have a very high chemisorption affinity for oxygen [39] and the presence of a thin insulating oxide layer can significantly impede the current flow across metal-semiconductor junctions.

The most common surface cleaning methods performed are chemical wet etches and plasma etches. Wet chemical treatments such as HF, buffered HF, HCl, HNO₃, H₂SO₄, KOH, NaOH, NH₄OH, (NH₄)₂S_x have been reported to reduce or modify the surface contamination [47]. Some of these substances such as HF, KOH and HCl are found to remove native Al and Ga oxides layers formed on AlGaN [35,39]. While HCl is more effective for removing oxides, HF is more effective for removing carbon and/or hydrocarbon contamination. Physical treatments, which enhance the device performance by lowering the surface N-vacancies and improving the surface stoichiometry include H₂-plasma, UV/O₃, O₂-plasma, N₂-plasma, Cl₂ based inductively coupled plasma (ICP) etching and Ar/N sputtering [47]. Most fabrication steps involve a series of surface cleaning steps to ensure the removal of contaminants from several contamination sources.

Contact angle measurement can serve as an indicator of foreign material present on the surface, while XPS and SIMS can provide detailed information on the molecular composition of the contaminants [48,49].

Surface analysis by XPS is accomplished by irradiating the sample with monoenergetic soft x-rays and analyzing the energy of the ejected electrons. Mg K α (1253.6 eV) or Al K α (1486.6 eV) x-rays are usually used. These photons have limited penetrating power in a solid, on the order of 1-10 μ m. They interact with atoms in the surface

region, causing electrons to be emitted by the photoelectric effect. The emitted electrons have measured kinetic energies given by:

$$\text{KE} = h\nu - \text{BE} - \phi_s \quad (2.44)$$

where $h\nu$ is the energy of the photon, BE is the binding energy of the atomic orbital from which the electron originates and ϕ_s is the spectrometer work function. Because each element has a unique set of binding energies, XPS can be used to identify and determine the concentration of the elements in the surface. Variations in the elemental binding energies arise from the differences in the chemical potential and polarizability of compounds. These chemical shifts can be used to identify the chemical state of the materials being analyzed. Argon ion etching is commonly used to obtain information on composition as a function of depth [50].

XPS measurements taken at different take off angles can be used to determine the thickness of a thin film present on the surface of another material by the following equation [51]:

$$d_{\text{film}} = \lambda_{\text{film}} \sin \alpha \ln [I_{\text{film}}/(\beta I_{\text{sub}}) + 1] \quad (2.45)$$

where d_{film} is the thickness of the thin film, λ_{film} is the photoelectron effective attenuation length in the thin film, α is the take off angle and I_{film} is the XPS peak intensity from the film, β is the film-to-substrate XPS photosensitivity ratio and I_{sub} is the XPS peak intensity from the substrate. This technique is referred to as an Angular-Resolved XPS (AR-XPS).

In SIMS the sample surface is sputtered by an ion beam and the emitted secondary ions are analyzed by a mass spectrometer. In static SIMS the primary ion dose is kept low ($\sim 10^{12}$ ions/cm²) to ensure that every ion hits a fresh surface. In dynamic SIMS systems multiple layers of atoms are removed in a typical rate of 0.5 to 5 nm/s. Most common ion sources for dynamic SIMS are O⁻, O₂⁺ and Cs⁺ ions. Oxygen

leads to the formation of oxides from which the original surface material emerges as a positively charged ion. Caesium is an efficient electron donor and mainly negatively charged secondary ions are generated. Dynamic SIMS typically use a magnetic mass analyzer, and for static SIMS the Time of Flight technique is most suitable [52].

Contact angle measurement can provide valuable information about the contaminants and oxide films on the surface. Contact angle is the angle at which the liquid interface meets the semiconductor interface. When purified water is used for contact angle measurement, surfaces which repel water (hydrophobic) yield high contact angles and surfaces that attract water (hydrophilic) yield small contact angles. Surface oxides formed on GaN or AlGaN layer were found to produce low contact angles. Carbon contamination, on the other hand, makes surfaces more hydrophobic [48, 53].

2.2 Literature Review on GaN and AlGaN/GaN Surface Treatment with Si Thin Films

Ion implantation of Si as a donor dopant in order to improve the tunnelling ohmic contacts to AlGaN and GaN materials and the associated diffusion and activation properties have been already well studied [4, 11, 13, 54–57]. However, due to the high temperature annealing required to activate the dopants (1,100°C) and to remove the ion induced crystal damage (1,300°C), a significant amount of attention has been given to alternative Si doping/incorporation techniques in III-N materials.

The use of Si on AlGaN/GaN epilayers in order to improve ohmic contacts or 2DEG densities can be classified into three categories. The first category of studies involves the deposition of Si-containing thin films on the AlGaN/GaN or GaN surface, followed by high temperature anneals and subsequent film removal. These

studies appear to conclude that solid state diffusion is responsible for the incorporation of significant amount of Si in the top epilayer. The second category includes the improvement of ohmic contacts with Si thin films deposited prior to metal evaporation and/or in between metal layers of ohmic stack. In this scheme Si takes a direct part in chemical reactions which occur during ohmic contact alloying. The third group of investigations involves improvement to 2DEG densities and/or surface passivation with Si films without intentional diffusion.

Pearson *et al.* [58] reported on substantial oxygen diffusion from SiO₂ film in p-type GaN in 1999, demonstrating that solid phase oxygen diffusion in GaN can take place at fairly low temperatures. SiO₂ layers were deposited on p-GaN by inductively coupled plasma chemical vapour deposition (ICP-CVD) using an ¹⁷O-enriched O₂ precursor. The samples were then annealed at 500 - 900°C for 5 minutes and the SiO₂ was removed. SIMS profiling appeared to show significant in-diffusion of ¹⁷O into the GaN under these conditions, with an incorporation depth of 180 nm after the 900 °C 5 minutes anneal. The ¹⁷O diffusion profiles indicated that the high dislocation density in the GaN strongly affected the effective penetration depth. Two diffusion fronts were identified and attributed to bulk diffusion and threading dislocation diffusion. The latter was assumed to dominate the diffusion process and the following diffusion coefficient and activation energy were obtained: $D_o = 4.5 \times 10^{-12} \text{ cm}^2 \text{ s}^{-1}$ and $E_{da} = 0.23 \text{ eV}$.

Lin *et al.* [59] has reported on a substantial reduction to specific contact resistivity on undoped GaN material using Ti/Al/Ni (15/120/30 nm) ohmic contact scheme using Si films. A 1.5 µm GaN epilayer surface was cleaned in Buffered Oxide Etchant (BOE) for 5 min. before evaporating Si. The sandwich structure consisted of a 50 nm thick Si metal layer deposited by e-beam evaporation system directly on the GaN, followed by a 100 nm SiO₂ cap layer deposited by photo-CVD. Samples were annealed in N₂ ambient for 2 hours in temperatures ranging from 825 to 1,000°C. After the heat

treatment a chemical solution consisting of HF, HNO₃, and CH₃COOH at a ratio of 6:20:7 was used to remove the Si and cap layers. SIMS profiling has shown a significant amount of Si in GaN epilayer: the depth of apparent diffusion of Si in the GaN films was found to be 150 nm after 1,000°C annealing temperature. No appropriate SIMS control samples were used in this work. The heat treatment decreased the specific contact resistivity; specifically the ρ_c values observed decreased from 3×10^{-5} to $5.6 \times 10^{-7} \Omega\text{-cm}^2$. The temperatures used for the heat treatments appeared to be significantly lower than seemingly required for a substantial diffusion to occur.

In 2005 Ishida *et al.* [60] reported on the use of Si deposition to achieve one order of magnitude improvement to Ti/Al ohmic contact specific resistivity on AlGaIn/GaN material. The contact area was covered by Si film while the rest of the area was covered by SiO₂ film, followed by the annealing in a nitrogen atmosphere. After removing the Si and SiO₂ film, Ti/Al ohmic electrodes were formed. The increase of the Si concentration near the surface area by the selective doping technique was apparently confirmed by SIMS, however, the appropriate control sample, which would have consisted of coated with Si AlGaIn surface (with no heat treatment), was not used (Fig. 2.19). The 30 min. annealing appeared to increase the Si concentration in the AlGaIn layer by one order of magnitude from $2 \times 10^{20} \text{ cm}^{-3}$ without any selective doping to $3 \times 10^{21} \text{ cm}^{-3}$ after the annealing. Si appeared to diffuse at least 10 nm into AlGaIn. Correspondingly, the specific contact resistivity was significantly reduced from $1 \times 10^{-5} \Omega\text{-cm}^2$ to $1.2 \times 10^{-6} \Omega\text{-cm}^2$. Although the information on anneal temperatures was not included in this paper, one might infer from the paper written by co-author Murata *et al.* [61] that a 1,000 °C 30 min long anneal was used. Even then this temperature is lower than one would expect for a significant diffusion of Si into AlGaIn.

In 2004 Pan *et al.* [62] reported on apparent Si diffusion in p-GaN. To perform the diffusion process, a 600 nm thick Si film was sputter deposited directly on the

Figure 2.19: SIMS ²⁸Si profiles taken from a Si doped and from non-doped samples [60].

p-GaN layer, and a 200 nm thick SiN_x cap layer was deposited on top by Plasma-Enhanced Chemical Vapour Deposition (PECVD). Annealing was performed in flowing N₂ atmosphere for 2 hours at temperatures from 800 to 1,100°C. After annealing, the SiN_x encapsulant and the Si film were removed selectively from GaN using an HF:HNO₃:CH₃COOH (6:20:7) solution. The ohmic metal stack consisted of Ti/Al (25/100 nm). SIMS depth profiling revealed an apparent diffusion depth of 20 and 50 nm for 1,000 and 1,100°C respectively. The diffusion coefficient and the diffusion activation energies were derived to be $D_o = 4 \times 10^{-7} \text{ cm}^2 \text{ s}^{-1}$ and $E_{da} = 2.2 \text{ eV}$. Diffusion depths for 800 and 900°C were found to be 3 and 10 nm respectively. However, no control samples for SIMS measurements were included in this work.

In a study performed by Jang *et al.* [63] solid phase diffusion of Si in GaN was analyzed for 700 - 1,000°C 2 hours long anneals and Si film thickness of 10 - 50 nm deposited by e-beam evaporation. Samples were encapsulated with 100 nm of SiO₂ or SiN_x deposited by PECVD. After the anneals, BOE etch was used to remove the SiO₂ and HF/HNO₃/CH₃COOH wet etch to remove the Si. SIMS profiling revealed an apparent diffusion depths of ~100 nm for 800-900°C. The diffusion coefficient

and the diffusion activation energies for Si diffusion were found to be $D_o = 2.07 \times 10^{-4} \text{ cm}^2 \text{ s}^{-1}$ and $E_{da} = 0.57 \text{ eV}$. The group concluded that the Si diffusion process lowered the specific contact resistivity of Ti/Al/Pt/Au contacts by about a factor of 2 from $1.3 \times 10^{-4} \text{ } \Omega\text{-cm}^2$ to $7 \times 10^{-5} \text{ } \Omega\text{-cm}^2$. The final value of specific contact resistivity was not strongly dependent on temperature in the range 750 - 900°C. Beyond a Si film thickness of 10 nm, there was also no strong dependence of the contact properties. SiO₂ capping was found to be superior to SiN_x and there was no surface roughening for anneal temperatures below 1,000°C. This study lacks an appropriate control sample for SIMS as well.

Selective doping of source and drain regions with silicon oxynitride (SiON) thin film in AlGaIn/GaN HEMTs was reported by Adivarahan and Khan *et al.* [64] in 2007. SiON layer was deposited using a PECVD system at a temperature of 200°C with silane, ammonia, nitrogen and oxygen as gas precursors and helium as a diluent. This was followed by 650°C anneal in a forming gas ambient for 5 min. Significant increase in the Si concentration in the barrier and the 2-D channel region (around 20 nm below the surface) was observed from SIMS profiling. The ohmic contact scheme was not reported but ohmic contact resistance was improved from 0.55 Ohm-mm to 0.23 Ohm-mm. 2DEG sheet resistance was improved from 310 Ohm/sq to 260 Ohm/sq. Given such a low anneal temperature and short anneal time, the solid phase diffusion of Si into AlGaIn is very questionable in this work.

The same selective doping technique using SiON was reported by Deng and Khan *et al.* [65] in 2007. The information on the film thickness and techniques used to remove it were not included in these papers.

Kim *et al.* [66] investigated Si/Ti ohmic contact to n-GaN in 2000. Two schemes, Ti(15 nm)/Si(50 nm)/Ti(40 nm)/Au(100 nm) and Ti(15 nm)/Si(146 nm)/Ti(40 nm)/Au(100 nm), were employed in the study. Best specific contact resistivities of $3.5 \times 10^{-5} \text{ } \Omega\text{-cm}^2$ and $7.4 \times 10^{-6} \text{ } \Omega\text{-cm}^2$ were obtained after 900°C ohmic metal

stepwise anneals for each scheme. Good ohmic contact resistivity was attributed to low work function of Ti-Si binary compounds: Ti_5Si_3 (3.69 - 3.73 eV), TiSi (3.94 - 3.99 eV) and TiSi_2 (3.95 - 4.18 eV). It was also found that current conduction in the ohmic contact was occurring via thermionic emission over a low barrier height.

In a study performed by Desmaris *et al.* [67] the ohmic metal stack was comprised of $\text{Si}(3/9 \text{ nm})/\text{Ti}/\text{Al}/\text{Ni}/\text{Au}$ were investigated using transmission electron microscopy and energy dispersive x-ray spectroscopy. The dependence of the contact resistance on the Si layer thickness and the temperature was correlated to the microstructure of the alloyed contacts. The reduction of the contact resistance by inserting a 3 nm thick Si layer under the $\text{Ti}/\text{Al}/\text{Ni}/\text{Au}$ metallization was attributed to diffusion of the Si into the AlGaIn layer. Increasing the Si thickness and/or the temperature resulted in the formation of Au-based silicides, which prevented the formation of low barrier TiN or AlN layers. Minimum ohmic contact resistance of 0.23 Ohm-mm was obtained for a 3 nm Si layer after 780°C anneal.

Mohammed *et al.* [68] reported on $\text{Ti}/\text{Si}(15/30 \text{ nm})/\text{Al}/\text{Si}(15/30 \text{ nm})/\text{Mo}/\text{Au}$ ohmic contact resistances to AlGaIn/GaIn heterostructures in 2006. Excellent ohmic contact formation with contact resistance and specific contact resistivity as low as 0.12 Ohm-mm and $3.8 \times 10^{-7} \Omega\text{-cm}^2$, respectively, have been obtained for 15 nm Si thickness scheme. The processing window for annealing temperatures was increased in comparison to ohmic metal scheme with no Si.

In 2008 Onojima *et al.* [69] investigated the improvements to 2DEG densities and to $\text{Ti}/\text{Al}/\text{Ni}/\text{Au}$ ohmic contact resistances with the use of a 2 nm Si layer on AlGaIn/GaIn heterostructure surface. The starting material structure was comprised of 24.5 nm thick $\text{Al}_{0.28}\text{Ga}_{0.72}\text{N}$ barrier layer, 2,000 nm thick GaIn layer and sapphire substrate. Si was deposited by e-beam evaporation. Ohmic contact resistance was improved from 0.96 Ohm-mm to 0.39 Ohm-mm. After the Si deposition, the 2DEG

density increased up to a level equivalent to that attained due to the Catalytic (Cat)-CVD SiN deposition ($0.8 \times 10^{13} \rightarrow 1.2 \times 10^{13} \text{ cm}^{-2}$). There was no difference between the sample deposited with a Si layer alone and the sample deposited with a Si/Cat-CVD SiN stacked layer. This indicated that the increase in the 2DEG density can be induced by Si atom deposition as well as SiN passivation, and more importantly, this increase in density was closely related to the interface formation between Si and AlGaIn. Hall mobility slightly decreased due to Si and Cat-CVD SiN deposition ($1.6 \times 10^3 \rightarrow 1.4 \times 10^3 \text{ cm}^2/(\text{Vs})$). Resulting 2DEG sheet resistance improved from 480 to 380 Ohms/sq. due to the Si deposition. AlGaIn/GaN HFETs with a length of a T-shaped gate of 150 nm exhibited good pinch-off behavior, and the maximum drain current densities were 1.2 A/mm for the device with Si treatment and 1.1 A/mm for that without Si. The 2 nm thick Si insert layer did not adversely affect the DC device characteristics. However, it can be seen from the drain DC I-V curves provided in the paper that the pinch off voltage changed from -11 V to -13 V on Si-coated samples. Even more drastic 2DEG sheet resistance improvement was observed by Onojima *et al.* [70] in the case of AlN(2 nm)/GaN structure (note that this is not a conventional AlGaIn/GaN heterostructure) where resistance was remarkably reduced from 60,356 to 388 Ohm/sq. by the deposition of a 2 nm thick Si layer on the AlN barrier surface. These results strongly support the “Si surface donor model”, in which Si atoms located at the SiN/AlGaIn interface act as positively ionized donors, which can partially neutralize the negative polarization charges of the AlGaIn surface and thereby increase the 2DEG density (Fig. 2.20).

Surface passivation effects were studied on AlGaIn/GaN HEMTs using SiO_2 , Si_3N_4 , and SiON formed by PECVD by Arulkumaran *et al.* [71] in 2004. This study demonstrated that other Si-containing passivation schemes, in addition to the widely used Si_3N_4 , can reduce current collapse and increase transconductance and peak current density.

Figure 2.20: Schematic model of the atomic arrangement at the SiN/AlGaIn interface [69].

The reviewed literature of GaN and AlGaIn/GaN surface treatment with Si thin films clearly demonstrates the importance of Si in GaN semiconductor materials. One of the greatest shortcomings of these studies, however, is the exclusion of other mechanisms by which the deposition of Si thin film, annealing and its removal can enhance the performance of ohmic contacts. Several other mechanisms, such as an incomplete Si film removal, chemical reaction on the topmost surface and roughening of the AlGaIn layer can be responsible for changes in electrical performance of fabricated devices. Samples containing as-deposited Si films were not included in SIMS profiling, thus the results from these studies can include forward scattering of surface Si atoms by primary ions. This artifact will exaggerate the diffusion depth of Si. Incomplete Si removal during wet etches will result in apparent Si diffusion during SIMS profiling and can enhance ohmic contacts as was demonstrated by studies using Si inserts and Si-containing ohmic metal stacks.

Previous research does not provide an answer to questions such as whether deep diffusion of Si is necessary to achieve better electrical results or whether interaction between the Si thin film and the topmost GaN or AlGaIn layer is sufficient. If high

dislocation density is important for this technique, would it work on modern GaN material with lower dislocation densities?

Also, 2DEG sheet resistance was shown to improve after surface passivation with Si thin films, but 2DEG sheet resistance after Si etch and subsequent HEMT fabrication was not reported. All the studies involving the diffusion of Si from Si films to improve ohmic contacts used anneals in excess of 30 minute and it is not clear whether a shorter RTA anneal time would be sufficient to obtain similar results.

Lastly, the effect of Si treatments followed by Si film removal on AlGaN surface wetting properties has not been reported. Surface cleaning has a profound effect on fabricated device properties and thus a change in hydrophobic properties of AlGaN after Si thermal diffusion could potentially be the reason behind better ohmic contacts.

Chapter 3

DESIGN AND FABRICATION

3.1 Process Flow Considerations and Starting Material

In order to study the effects and physical mechanism behind electrical enhancement in AlGa_N/Ga_N devices after surface “treatments” and “coatings” with Si-containing thin films, three fabrication runs were developed and implemented. “Treatment” is defined as the e-beam deposition of Si-containing thin film, encapsulation with SiO₂ deposited using a PECVD, RTA annealing and the films removal with a wet etch. “Coating” is defined as the e-beam deposition of Si film on the AlGa_N/Ga_N surface. The three sets of Ga_N samples as well as the associated process runs were named “Batch-1”, “Batch-2” and “Batch-3”. The main goal of Batch-1 fabrication run was to confirm the changes in electrical performance on treated AlGa_N/Ga_N surfaces with Si and SiON thin films. The 2DEG electron mobility and density were monitored during stepwise RTA annealing and the performance of fabricated test structures and devices was compared.

Batch-2 and Batch-3 fabrication runs built on the results from Batch-1 and also included samples permanently coated with Si film. The analysis and characterization

of these samples were aided by SIMS profiling, XPS chemical composition and water contact angle measurements in order to investigate further the effect of Si treatments and coatings on the AlGaN surface.

Starting material for all three batches consisted of rectangular pieces diced from a high quality $\text{Al}_{0.28}\text{Ga}_{0.72}\text{N}/\text{AlN}/\text{GaN}/\text{GaN:Fe}$ (20/1/200/1,800 nm) on SiC substrate wafer purchased from IQE corporation. Nominal sheet resistance R_s for this passivated material is ~ 300 Ohm/sq, 2DEG electron mobility μ is $2,000 \text{ cm}^2/(\text{Vs})$ and 2DEG sheet electron concentration n_s is $1.0 \times 10^{13} \text{ cm}^{-2}$. Each sample came from a different portion of the wafer thus some scatter in the results was expected between identically treated samples. The wafer structure and fabricated HEMT structure are shown together in Fig. 2.4.

Amorphous silicon (a-Si) was deposited with a Balzers BA510 e-beam evaporation system using a high resistivity (1,500 Ohm-cm) electronic grade n-type (doped with phosphorous) Si wafer as a source. Silicon oxynitride was deposited in an Oxford PECVD system 100, using silane diluted in helium, ammonia, nitrous oxide and nitrogen. Silicon dioxide (SiO_2) was deposited with a Plasma-Therm 700 PECVD system using silane dilute in helium and nitrous oxide.

Existing experimental lithographic photo-masks were used to define device features using a Karl-Suss MJB-3 mask aligner with a 365 nm Hg i-line source. The mask layout can be seen in Fig. 3.1. An Agilent 4155C semiconductor parameter analyzer and software controlled digital multimeter were used for electrical measurements of fabricated devices and test structures. HEMTs with the gate length of $1 \mu\text{m}$, drain-to-gate spacing of $4 \mu\text{m}$ and gate-to-source spacing of $1.5 \mu\text{m}$ were used for electrical measurements in this work.

The following sections describe the development of process flows for each of these batches. A detailed step-by-step fabrication process flow for each run as well as an outline of NRC's baseline GaN technology can be found in Appendix A.

Figure 3.1: Lithographic photo-mask. TLM, c-TLM and HEMT structures are probeable after ohmic metal anneal.

3.2 Batch-1 Process Flow Design and Measurements

The focus of the first experimental run was to determine whether there were any observable electrical changes in 2DEG density and HEMT performance with Si surface treatments. The following two sources of Si were of most interest, since both of them were reported in the literature: Si and SiON. The e-beam evaporation method for Si deposition was preferred, since the ohmic contact metal stack utilized the same deposition technique and future *in-situ* deposition was possible. At the time of this work, however, it was not possible to evaporate both Si and ohmic contact metals in the same evaporation system due to equipment contamination concerns. Target samples were kept at low temperature during the evaporation to avoid crystallization

and thereby obtain a-Si films [72]. However, it was expected that the a-Si would crystallize during subsequent high temperature anneals. Since neither the content of oxygen and nitrogen nor the refractive index of SiON thin film treatments has been reported in the literature, a 1:1 ratio of SiO₂ to Si₃N₄ was used as a starting point.

The a-Si thin film thickness was chosen to be 30 nm, sufficiently thicker than the critical 10 nm thickness established by Jang *et al.* [63]. Magnets were used on corners of the samples to keep them in place during e-beam evaporation, thus the corners were free of a-Si. Since it was not clear whether SiON can contribute more or less Si atoms for possible diffusion than pure Si, the silicon oxynitride thickness was set to be twice the value of a-Si: 60 nm.

A SiO₂ encapsulation layer was used by several groups in order to protect the AlGaIn surface from degradation caused by high temperature anneals. It was found to be superior to the SiN_x encapsulation scheme by Jang *et al.* as well. Thus the encapsulation layer used here was comprised of 100 nm SiO₂, deposited by PECVD.

In order to establish whether high temperature annealing alone could produce observable electrical changes in AlGaIn/GaN heterostructures, RTA anneals were performed on two untreated samples. The samples were selectively stepwise RTA annealed in N₂ flow starting at 400°C and ending at 950°C with Hall measurements of 2DEG density and electron mobility performed between each anneal. Anneal times ranged from 1 minute to 8 minutes for high and low anneal temperatures respectively. Samples coated with thin films had the corners wet etched in BOE for Hall system probes access. In order to reduce the error in Hall measurement, each sample was measured in four orientations achieved by 90 degree rotations. Hall measurement results provided a feedback on the requirement for further anneals for each sample.

Following the anneals, both the SiO₂ cap and SiON were removed in BOE. Etch times were determined using Si test wafers included in the PECVD deposition. HF:HNO₃ mixtures etch Si at rates exceeding 100 nm/s for a wide range of

compositions [73]. Therefore, Si films were etched from the Si-treated samples using HF:HNO₃ 1:3 mixture for 5-10 seconds.

Standard NRC processing steps were followed to implement ohmic contacts, isolation mesa, Schottky gates, silicon nitride passivation and metal-1 interconnects.

The following samples were used in this run:

- **A1:** a control sample.
- **B1 and B2:** samples RTA annealed to 850°C and 900°C respectively.
- **S1 and S2:** samples coated with a-Si, encapsulated with SiO₂, RTA annealed to 950°C and 900°C respectively. The encapsulation and Si films were removed with the BOE and HF:HNO₃ etch.
- **O1 and O2:** samples coated with SiON, encapsulated with SiO₂, RTA annealed to 850°C and 900°C respectively. The encapsulation and SiON films were removed with the BOE etch.

The following electrical device measurements were performed after the specified fabrication steps:

- **Ohmic Metal RTA Anneal:** ohmic contact and sheet resistance from c-TLM structures.
- **Gate metal anneal:** ohmic contact and sheet resistance from c-TML and TLM structures, DC and pulsed HEMT I-V characteristics - current collapse, gate Schottky junction I-V characteristics, HEMT drain-source breakdown voltage.
- **Passivation and Via1 etch:** ohmic contact and sheet resistance from c-TML and TLM, DC and pulsed HEMT I-V characteristics - current collapse, gate Schottky junction characteristics, HEMT drain-source breakdown voltage.
- **1ME deposition:** ohmic contact and sheet resistance from c-TML and TLM.

The following table (Table 3.1) provides the details on critical process steps as well as AlGaIn/GaN material measurements for Batch-1.

Batch-1 Samples	A1	B1	B2	S1	S2	O1	O2
Hall Measurement(1)	×	✓	✓	✓	✓	✓	✓
E-beam a-Si (30 nm)	×	×	×	✓	✓	×	×
PECVD SiON (60 nm)	×	×	×	×	×	✓	✓
PECVD SiO ₂ (100 nm)	×	×	×	✓	✓	✓	✓
Hall Measurement(2)	×	✓	✓	✓	✓	✓	✓
RTA Anneals							
—400°C 4 min/Hall(3)	×	✓	✓	✓	✓	✓	✓
—650°C I 4 min/Hall(4)	×	✓	✓	✓	✓	✓	✓
—650°C II 4 min/Hall(5)	×	✓	✓	✓	✓	✓	✓
—700°C 8 min/Hall(6)	×	✓	✓	✓	✓	✓	✓
—750°C 2x 4 min/Hall(7)	×	✓	✓	✓	✓	✓	✓
—800°C 2x 2 min/Hall(8)	×	✓	✓	✓	✓	✓	✓
—850°C 2x 1 min/Hall(9)	×	✓	✓	✓	✓	✓	✓
—875°C 2x 1 min/Hall(10)	×	×	✓	×	✓	×	✓
—900°C I 2x 1 min/Hall(11)	×	×	✓	×	✓	×	✓
—900°C II 6x 1 min/Hall(12)	×	×	✓	×	✓	×	✓
—925°C 1 min/Hall(13)	×	×	×	✓	×	×	×
—950°C 1 min/Hall(14)	×	×	×	✓	×	×	×
BOE Etch	×	×	×	✓	✓	✓	✓
HF:HNO ₃ 1:3 5 sec Etch	×	×	×	✓	✓	×	×
Hall Measurement(15)	×	✓	✓	✓	✓	✓	✓
Ohmic RTA 845°C 1 min	✓	✓	✓	✓	✓	✓	✓

Table 3.1: Batch-1 samples and the details on corresponding critical process steps and material measurements. The number in brackets indicates the Hall measurement label.

3.3 Batch-2 Process Flow Design and Measurements

Batch-2 was designed to build on the results from the Batch-1 experiment. Due to promising results from the treatments with a-Si films, Batch-2 experiment focused on treatments from this source during one-time RTA anneals at temperatures of 850°C for 15 minutes, 950°C for 5 minutes and 1,000°C for 5 minutes. The same 30 nm e-beam evaporated a-Si and 100 nm PECVD deposited SiO₂ films were used.

The question was raised whether the HF:HNO₃ wet etch alone could produce the electrical changes which were discovered on a-Si treated samples from Batch-1. Therefore, an additional sample was included with the purpose of determining the effect of this step on 2DEG densities (measured by Hall effect system before and after the wet etch) as well as on the performance of fabricated c-TLM/TLM structures and HEMTs.

In order to study the difference between Si-treated and Si-coated samples, two samples with 5 nm a-Si layer were included in the run. Reviewed literature indicated that the thickness of the coating layer should be around 2-3 nm. However, it was not possible to deposit Si *in-situ* prior to ohmic metal evaporation and thus native oxide formation which could reduce the thickness of the Si film was a concern. Si wafers stored for 1 week time period in air form a native oxide of 1.6 to 2 nm thick, weakly depending on initial surface cleaning [74]. Thus, in order to account for the native oxide growth, the a-Si thickness was chosen to be 5 nm deposited by e-beam evaporation.

For the investigation of the artifact produced during SIMS profiling as well as for XPS referencing, two additional samples with 5 nm of e-beam deposited a-Si were included.

It was anticipated that ohmic metal would require different alloying temperatures

for each sample to reach optimum results, thus samples were selectively stepwise RTA annealed in N₂ flow starting at 750°C and ending at 845°C with c-TLM measurements taken in between each anneal.

Ohmic metal got heavily oxidized due to a glitch in RTA system's control software. The ohmic contact resistance results from this batch became suspect. Further processing of this batch was halted following the ohmic metal RTA anneals. Batch-3 run was designed to replicate in parts the experiments planned to be performed on Batch-2 samples.

XPS (using Al K α 1486.6 eV source) and dynamic SIMS (using Cs source) measurements were performed on reference and processed samples. Ar sputtering was employed for XPS measurement on the reference sample to reveal simultaneously the location of Si and Ga peaks where the former was used for charge referencing. Carbon peak referencing in XPS was shown to be unreliable for high temperature treated samples [75]. The elemental Si 2*p* peak was assumed to have a location of 99.3 eV to calibrate the energy axis. The "XPSPeak v4.1.1" software was used to fit the spectra in order to obtain the position and the intensity of each peak.

The following samples were used in this run:

- **B1:** a control sample.
- **A1:** HF:HNO₃ treated sample.
- **S1 and S2:** samples coated with a-Si, RTA annealed at 950°C for 5 min. The encapsulation and Si films were removed with the BOE and HF:HNO₃ etch.
- **L1:** a sample coated with a-Si, RTA annealed at 850°C for 15 min. The encapsulation and Si films were removed with the BOE and HF:HNO₃ etch.
- **H1:** a sample coated with a-Si, RTA annealed at 1,000°C for 5 min. The encapsulation and Si films were removed with the BOE and HF:HNO₃ etch.
- **C1 and C2:** samples coated with 5 nm of a-Si where the a-Si film stayed in place.
- **R1 and R2:** XPS and SIMS reference samples coated with 5 nm of a-Si.

The following electrical device measurements were performed after specified fabrication steps:

- **Ohmic Metal RTA Anneal:** ohmic contact and sheet resistance from c-TLM, saturation current density over 2 μm c-TLM gap.

Table 3.2 provides the details on critical process steps for Batch-2 as well as AlGaIn/GaN material measurements.

Batch-2 Samples	B1	A1	S1	S2	L1	H1	C1	C2	R1	R2
E-beam a-Si (30 nm)	×	×	✓	✓	✓	✓	×	×	×	×
E-beam a-Si (5 nm)	×	×	×	×	×	×	✓	✓	✓	✓
PECVD SiO ₂ (100 nm)	×	×	✓	✓	✓	✓	×	×	×	×
Hall Measurement(1)	×	✓	×	×	×	×	×	×	×	×
Thin Film RTA Anneals										
—850°C 15 min	×	×	×	×	✓	×	×	×	×	×
—950°C 5 min	×	×	✓	✓	×	×	×	×	×	×
—1,000°C 5 min	×	×	×	×	×	✓	×	×	×	×
BOE Etch	×	×	✓	✓	✓	✓	×	×	×	×
HF:HNO ₃ 1:3 10 sec	×	✓	✓	✓	✓	✓	×	×	×	×
Hall Measurement(15)	×	✓	✓	✓	✓	✓	×	×	×	×
Ohm. Cont. RTA Anneals										
—750/800/825/835°C 1 min	✓	✓	✓	✓	✓	✓	✓	✓	×	×
—840°C 1 min	✓	✓	✓	✓	✓	✓	✓	✓	×	×
—845°C 1 min	×	×	×	×	×	×	✓	✓	×	×
SIMS Profiling	✓	×	✓	×	✓	✓	×	×	✓	×
XPS Surface Measurement	×	✓	✓	✓	×	×	×	×	×	✓
XPS with Ar Sputtering	×	×	×	×	×	×	×	×	×	✓

Table 3.2: Batch-2 samples and the details on corresponding critical process steps and material measurements. The number in brackets indicates the Hall measurement label.

3.4 Batch-3 Process Flow Design and Measurements

Batch-3 was designed to replicate the Batch-2 experiment with additional XPS as well as water contact angle measurements. Two samples were dedicated to each category: control, Si-treated, and Si-coated samples.

The treated samples were coated with a 30 nm e-beam evaporated a-Si thin films and encapsulated with a 100 nm PECVD deposited SiO_2 . Both samples were RTA annealed at 950°C for 5 min, followed by BOE and HF:HNO_3 etch. The a-Si film thickness of 5 nm was used for the coated samples. XPS surface chemical composition measurement was performed on Si-treated samples.

In order to establish and compare the difference in hydrophobic properties of the AlGaIn surface, contact angle measurements were performed after each processing step leading up to ohmic metal deposition, such as the developing of photo-resist and final cleans (these steps had to be repeated prior to the actual ohmic metal deposition). This was followed by AR-XPS surface chemical analysis measurement at take off angles of 30, 45 and 77.5° in order to establish the thickness and the composition of the remnant Si-containing film.

After ohmic metal deposition, the samples were selectively stepwise RTA annealed in N_2 flow starting at 825°C and ending at 835°C with c-TLM measurements performed in between.

Standard NRC processing steps were followed to implement an isolation mesa, Schottky gates, silicon nitride passivation and metal-1 interconnects.

The following samples were used in this run:

- **B1 and B2:** control samples.
- **S1 and S2:** samples coated with a-Si, RTA annealed at 950°C for 5 min. The encapsulation and Si films were removed with the BOE and HF:HNO_3 etch.

- **C1 and C2:** samples coated with 5 nm of a-Si where the a-Si film stayed in place.

The following electrical device measurements were performed after specified fabrication steps.

- **Ohmic Metal RTA Anneal:** ohmic contact and sheet resistance from c-TLM, saturation current density over 2 μm c-TLM gap.
- **Gate metal anneal:** ohmic contact and sheet resistance from TLM, saturation current density over 2 μm c-TLM gap, DC and pulsed HEMT I-V characteristics - current collapse, HEMT transconductance and threshold voltage.
- **1ME deposition:** ohmic contact and sheet resistance from c-TML and TLM, saturation current densities over 2 μm c-TLM gap, DC and pulsed HEMT I-V characteristics - current collapse, gate Schottky junction I-V characteristics, HEMT transconductance and threshold voltage, gate leakage current versus drain voltage for standard and field plate assisted HEMTs.

Table 3.3 provides the details on critical process steps for Batch-3.

Batch-3 Samples	B1	B2	S1	S2	C1	C2
E-beam a-Si (30 nm)	×	×	✓	✓	×	×
E-beam a-Si (5 nm)	×	×	×	×	✓	✓
PECVD SiO ₂ (100 nm)	×	×	✓	✓	×	×
Thin Film RTA Anneals						
—950°C 5 min	×	×	✓	✓	×	×
BOE Etch	×	×	✓	✓	×	×
HF:HNO ₃ 1:3 10 sec	×	×	✓	✓	×	×
XPS Surface Measurement	×	×	✓	✓	×	×
Contact Angle Measurement						
—Initial	✓	✓	✓	✓	✓	✓
—Ace/IPA/DI	✓	✓	✓	✓	✓	✓
—Ace/IPA Scrub/DI	✓	✓	✓	✓	✓	✓
—HMDS Prime	✓	✓	✓	✓	×	✓
—Resist Coat/Bake/Develop	✓	✓	✓	✓	×	✓
—Microwave Descum	✓	✓	✓	✓	×	✓
—Ohmic Pre Clean/DI Rinse	✓	✓	✓	✓	×	✓
—2nd DI Rinse	✓	✓	✓	✓	×	✓
—20 hours storage	✓	✓	✓	✓	×	✓
AR-XPS 30°/45°/77.5°	×	×	×	✓	×	×
Ohm. Cont. RTA Anneals						
—825/830°C 1 min	✓	✓	✓	✓	✓	✓
—835°C 1 min	×	×	×	✓	✓	×

Table 3.3: Batch-3 samples and the details on corresponding critical process steps and material measurements.

Chapter 4

RESULTS

4.1 Introduction

The results of the electrical and chemical measurements are presented in this chapter. Since the main goal of Si treatments and coatings was to improve the electrical performance of fabricated devices and because the electrical measurements for the most part were performed prior to the the surface composition measurements, they are presented in the beginning of the chapter. The 2DEG properties measured by Hall effect are given first, followed by the electrical measurement results of fabricated TLM structures and transistors. The chapter is concluded with the most important and telling results obtained from the surface chemical composition measurements obtained by SIMS and XPS as well as with the water contact angle measurement on the AlGaN surface. Control samples processed by conventional means are marked with a ‘†’ throughout this chapter for easier identification.

4.2 2DEG Sheet Resistance, Electron Density and Mobility by Hall

4.2.1 Batch-1

The results of Hall measurements performed on Batch-1 samples are organized in three groups of plots. Fig. 4.1 shows the evolution of the 2DEG sheet resistance as a function of processing and RTA anneal steps (Table 3.1). Fig. 4.2 and Fig. 4.3 present the evolution of 2DEG electron mobility and density.

There is a clear advantageous trend in μ , n_s and thus R_s on Si-treated samples S1 and S2 before Si etch. Following the removal of the films, S1 and S2 displayed an increase in the 2DEG carrier mobility and a decrease in the 2DEG sheet carrier density. This resulted in a modest increase in R_s , however, the final value remained very low, 310 and 320 Ohm/sq for samples S1 and S2 respectively. The final mobilities on samples S1, S2, O1 and O2 are comparable and are better than on RTA annealed samples B1 and B2.

(a) B1, RTA annealed control

(b) B2, RTA annealed control

(c) S1, a-Si film

(d) S2, a-Si film

(e) O1, SiON film

(f) O2, SiON film

Figure 4.1: Batch-1 evolution of the 2DEG sheet resistance R_s as a function of processing steps. The spread associated with each data point is $\sim 5\text{-}10$ Ohm/sq. Step 1 and 2: before and after the deposition of the thin films. Steps 3-14: RTA anneals. Step 15: after the HF:HNO₃ wet etch.

Figure 4.2: Batch-1 evolution of the 2DEG electron mobility μ as a function of processing steps. Step 1 and 2: before and after the deposition of the thin films. Steps 3-14: RTA anneals. Step 15: after the HF:HNO₃ wet etch.

(a) B1, RTA annealed control

(b) B2, RTA annealed control

(c) S1, a-Si film

(d) S2, a-Si film

(e) O1, SiON film

(f) O2, SiON film

Figure 4.3: Batch-1 evolution of the 2DEG electron density n_s as a function of processing steps. Step 1 and 2: before and after the deposition of the thin films. Steps 3-14: RTA anneals. Step 15: after the HF/HNO₃ wet etch.

4.2.2 Batch-2

Table 4.1 summarizes the Hall system measurements of Batch-2 samples. An improvement to the 2DEG sheet resistance was seen on sample A1 following the HF:HNO₃ wet etch. However, this improvement did not materialize into better sheet resistances in TLM structures, as shown further in this chapter. This result was not unexpected, since unpassivated AlGaIn/GaN surfaces are very sensitive to surface conditions. Very low sheet resistances were obtained on all samples except S1.

	Before the Thin Film Dep.			After HF:HNO ₃ Wet Etch		
	R_s	μ	n_s	R_s	μ	n_s
	Ohm/sq	cm ² /(Vs)	cm ⁻²	Ohm/sq	cm ² /(Vs)	cm ⁻²
A1	326	2020	9.49×10^{12}	311	2040	9.86×10^{12}
S1				339	1950	9.42×10^{12}
S2				300	2000	1.04×10^{13}
L1				317	1950	1.01×10^{13}
H1				302	1990	1.04×10^{13}

Table 4.1: Batch-2 2DEG sheet resistance, electron mobility and density before the deposition of the a-Si thin film and after the HF:HNO₃ wet etch.

4.3 Ohmic Contacts and 2DEG Sheet Resistance from c-TLM and TLM Structures

4.3.1 Batch-1

Table 4.2 summarizes Batch-1 TLM and c-TLM measurement results at several stages of the process flow. Superior to other samples ohmic contact and 2DEG sheet resistances were obtained on Si-treated samples S1 and S2. However, from the control sample measurements, as well as Batch-2 and Batch-3 annealing conditions, it was established that the ohmic metal in Batch-1 was over-annealed. Better 2DEG sheet resistance on Si-treated samples after the passivation step, which normally improves the 2DEG sheet resistance, may indicate that the Si treatments provide additive passivating effects. The variation between the c-TLM and TLM measurements represents the scatter of the data across the sample since the two patterns were located apart from each other.

	RTA anneal		Gate Anneal		Passivation		1ME Dep.	
	R_c	R_s	R_c	R_s	R_c	R_s	R_c	R_s
	Ohm-mm	Ohm/sq	Ohm-mm	Ohm/sq	Ohm-mm	Ohm/sq	Ohm-mm	Ohm/sq
[†] A1 c-TLM	1 65	322	-	-	1 60	325	1 63	323
[†] A1 TLM	-	-	1 86	319	1 80	314	1 68	314
B1 c-TLM	1 12	437	-	-	1 12	313	1 07	300
B1 TLM	-	-	1 29	307	1 23	304	1 20	304
B2 c-TLM	1 38	503	-	-	1 14	340	1 19	333
B2 TLM	-	-	1 45	323	1 39	320	1 29	320
S1 c-TLM	0 85	360	-	-	0 84	293	0 83	283
S1 TLM	-	-	0 93	307	0 9	289	0 82	286
S2 c-TLM	0 85	370	-	-	0 78	298	0 83	290
S2 TLM	-	-	1 02	317	0 93	291	0 84	290
O1 c-TLM	1 30	463	-	-	1 24	316	1 28	309
O1 TLM	-	-	1 41	311	1 38	305	1 26	304
O2 c-TLM	1 40	455	-	-	-	-	-	-
O2 TLM	-	-	1 44	314	-	-	-	-

Table 4.2: Batch-1 ohmic contact and 2DEG sheet resistances from a four point probe measurement of c-TLM and TLM structures. Averages of two to four measurements were taken for each sample. 1ME TLM results combine the measurements of ohmic-probeable as well as of 100x64 and 100x32 μm 1ME-probeable TLMs.

Comparison bar plots of the final ohmic contact and 2DEG sheet resistances obtained after passivation and 1ME deposition step in Batch-1 are given in the figure below (Fig. 4.4).

(a) Ohmic contact resistance (R_c) from c-TLM (red) and TLM (green)

(b) 2DEG sheet resistance (R_s) from c-TLM (red) and TLM (green)

Figure 4.4: Batch-1 ohmic contact and 2DEG sheet resistances after passivation and 1ME deposition steps.

4.3.2 Batch-2

Ohmic contact and 2DEG sheet resistances as well as the saturation current density over 2 μm c-TLM gap at 9.9 V bias after RTA anneals at temperatures of 825, 835, 840 and 845°C are summarized in Table 4.3. Further processing of this batch was halted due to the heavy oxidation of the ohmic metal stack. The metal on the samples has become discoloured, however, the electrical probing could still be successfully performed. It should be noted that this accidental oxidation affected the ohmic metal, which is known to be easily oxidized. However, the temperature of 800°C and low oxygen content inside the RTA system were not sufficient to alter significantly the composition of treated, coated and untreated AlGaIn layers. The electrical results, however, are questionable given the oxidation of the metal.

All Si-treated and Si-coated samples demonstrated far superior characteristics than the control sample [†]B1 and the acid treated sample A1 under these circumstances. Optimum alloying temperatures for Si-treated samples (S1, S2, L1, H1) and for Si-coated samples (C1, C2) were shifted to higher values. Si-treated and coated samples exhibited significantly higher saturation current densities over 2 μm c-TLM gap, the measurement which relies both on the performance of the ohmic contacts and the sheet resistance. In addition, low (<1 Ohm-mm) contact resistance values were achieved more consistently on the treated and coated samples. It can be concluded that the Si treatments and coatings can improve the ohmic contacts and device current densities in unfavourable processing conditions, such as the presence of oxygen during the annealing of ohmic metal.

The pre-passivating properties of Si treatments and coatings are seen in the 2DEG sheet resistance values on the control sample, acid treated sample and Si-treated/coated samples (Fig. 4.5). The values were taken for RTA anneals producing

	825°C RTA		835°C RTA			840°C RTA			845°C RTA		
	R_c	R_s	R_c	R_s	I_d	R_c	R_s	I_d	R_c	R_s	I_d
	Ohm-mm	Ohm/sq	Ohm-mm	Ohm/sq	mA/mm	Ohm-mm	Ohm/sq	mA/mm	Ohm-mm	Ohm/sq	mA/mm
[†] B1	0.91	398	2.37*	400*	500	1.06*	385*	637	-	-	-
A1	1.18	461	2.75*	425*	388	2.00*	425*	338	-	-	-
S1	2.27*	287*	0.68	305	1040	0.68	345	818	-	-	-
S2	6.25*	300*	0.74	293	1130	0.73	305	900	-	-	-
L1	5.50*	285*	0.69	325	1004	0.60	326	944	-	-	-
H1	3.08*	405*	0.58	300	1204	0.55	323	1067	-	-	-
C1	9.38*	638*	0.74	320	1149	0.45	330	1188	0.65	315	1146
C2	-	-	-	-	-	0.46	318	1020	0.60	318	1014

Table 4.3: Batch-2 ohmic contact, 2DEG sheet resistances and current density over 2 μm c-TLM gap from a four point probe measurement of c-TLM and TLM structures during RTA annealing. Averages of two to four measurements were taken for each sample. ‘*’ denotes non-linearity in TLM data.

best ohmic contacts based on both ohmic contact resistance values and the saturation current densities. An exception was made for samples [†]B1 and A1, where the lowest 2DEG sheet resistances were used. The fact that Si-treated and coated material showed signs of passivation after the ohmic metal RTA anneals but before the SiN passivation is very beneficial from the standpoint of in-process device performance monitoring, where process adjustments (such as additional ohmic contact RTA anneals) can still be made.

Figure 4.5: Batch-2 2DEG sheet resistance values (R_s) after ohmic contact RTA anneals. 835°C results were used for Si-treated samples. 840°C results were used for †B1, A1 and Si-coated samples. Pre-passivating effects of Si treatments and coatings can be clearly seen.

4.3.3 Batch-3

Ohmic contact and 2DEG sheet resistances as well as the saturation current density over 2 μm c-TLM gap at 9.9 V bias after RTA anneals at temperatures of 825, 830 and 835°C are summarized in Table 4.4. The devices on Si-treated (S1, S2) and coated (C1, C2) samples have shown superior to the control samples saturation current densities at this stage, even in the absence of silicon nitride passivation films. The ohmic contact resistance values were comparable on all samples.

	825°C RTA			830°C RTA			835°C RTA		
	R_c	R_s	I_d	R_c	R_s	I_d	R_c	R_s	I_d
	Ohm-mm	Ohm/sq	mA/mm	Ohm-mm	Ohm/sq	mA/mm	Ohm-mm	Ohm/sq	mA/mm
†B1	1.18	380	423	0.68	387	917	-	-	-
†B2	1.55	390	344	0.60	396	776	-	-	-
S1	4.00*	365*	490	0.53	352	1114	-	-	-
S2	3.00*	400*	535	0.45*	340*	809	0.68	292	1073
C1	-	-	-	2.00*	300*	645	0.59	266	1353
C2	3.65*	330*	532	0.58	290	1141	-	-	-

Table 4.4: Batch-3 ohmic contact, 2DEG sheet resistances and current density over 2 μm c-TLM gap from a four point probe measurement of c-TLM and TLM structures during RTA annealing. Averages of two to four measurements were taken for each sample. ‘*’ denotes non-linearity in TLM data.

Figure 4.6 demonstrates again the pre-passivating effect from the Si treatment and coating on the 2DEG sheet resistances in AlGaIn/GaN heterostructure in this batch. Since the contact resistance values were comparable across all samples and there were no accidental ohmic metal oxidation or over annealing, it was confirmed that this pre-passivation indeed occurred as a result of Si treatments and coatings.

Figure 4.6: Batch-3 final 2DEG sheet resistance values (R_s) after ohmic contact RTA anneals. Pre-passivating effects of Si treatments and coatings can be clearly seen.

Table 4.5 summarizes Batch-3 TLM measurements after gate anneal and 1ME deposition processing steps. The Si-coated samples C1 and C2 demonstrated superior sheet resistances and current densities. These characteristics were further improved by the passivation. Samples S1 and S2 were likely affected by the processing steps performed for the water contact angle measurement during this fabrication run and thus did not exhibit high current densities and low sheet resistances before passivation with SiN. The Si treatment and coating did not produce any improvements to ohmic contacts.

	Gate Anneal			Passivation and 1ME Dep.		
	R_c	R_s	I_d	R_c	R_s	I_d
	Ohm-mm	Ohm/sq	mA/mm	Ohm-mm	Ohm/sq	mA/mm
[†] B1 c-TLM	-	-	924	0.57	283	1053
[†] B1 TLM	0.85	368	-	0.53	287	-
[†] B2 c-TLM	-	-	837	0.61	295	945
[†] B2 TLM	1.02	386	-	0.53	287	-
S1 c-TLM	-	-	784	0.55	288	1084
S1 TLM	0.82	401	-	0.61	279	-
S2 c-TLM	-	-	680	0.55	307	872
S2 TLM	1.28	306	-	0.76	287	-
C1 c-TLM	-	-	1037	0.56	278	1180
C1 TLM	0.76	275	-	0.66	271	-
C2 c-TLM	-	-	1050	0.54	276	1296
C2 TLM	0.76	291	-	0.52	285	-

Table 4.5: Batch-3 ohmic contact, 2DEG sheet resistances and current density over 2 μm c-TLM gap from a four point probe measurement of c-TLM and TLM structures. Averages of two to six measurements were taken for each sample. 1ME TLM results combine the measurements of ohmic-probeable as well as of 100x64 and 100x32 μm 1ME-probeable TLMs.

The final ohmic contact and 2DEG sheet resistance values after the passivation and 1ME deposition are plotted in Fig. 4.7. It is evident from this figure that under the right processing conditions the control, Si-treated and Si-coated samples had comparable ohmic contact resistances.

(a) Ohmic contact resistance (R_c) from c-TLM (red) and TLM (green)

(b) 2DEG sheet resistance (R_s) from c-TLM (red) and TLM (green)

Figure 4.7: Batch-3 ohmic contact and 2DEG sheet resistances after passivation and 1ME deposition steps.

Since the ohmic contacts on the control samples showed good resistance values after the anneals at 830°C and 835°C in Batch-3, it was concluded that the samples in Batch-1 were generally over annealed after 845°C RTA anneal. Because of heavy oxidation of the ohmic metal in Batch-2, only Batch-3 ohmic contact and 2DEG sheet resistance results were completely reliable. Thus the conclusions on the ohmic contact resistance values were taken from the results of this batch. The Si treatment and coatings were found to not affect the final values of the ohmic contact resistances significantly. The values remained in the range of 0.5 to 0.6 Ohm-mm. It is worth pointing out, however, that NRC's processing of ohmic contacts was already highly optimized. Based on Batch-1 and Batch-2 results, Si treatments and coatings can positively influence the ohmic contact resistances in adverse (or poorly optimized) processing conditions. There is a clear evidence of beneficial pre-passivation effects due to Si treatment and coating exhibited in Batch-2 and Batch-3, while Batch-1 Si-treated samples displayed lower 2DEG sheet resistance than on control samples even after the passivation step.

4.4 Ohmic Contact Surface Morphology by Visual Inspection and Surface Profiling

Batch-1 Si-treated samples S1 and S2 displayed a slightly better surface morphology and a similar line edge definition compared to the control and SiON-treated samples. The edge definition was worse on Si-coated samples in Batch-3. Batch-2 and Batch-3 Si-treated and Si-coated samples developed cracks in the ohmic metal stack following ohmic contact RTA anneals. The cracks in the ohmic metal were not noticed on Batch-1 samples. These defects created noticeable craters in 1ME overlayer upon further RTA annealing in Batch-3. The effect was most prominent on Si-coated samples (Fig. 4.8). It is clear that this phenomena is related to photoresist trapped in the cracks in the films which is not fully developed off, and thus results in a small explosion in the 1ME interconnect during annealing. Surface profiling with Dektak revealed a distinctive rise in the surface around the cracks (300 - 500 nm) and on the edges of the craters (1.0 - 2.0 μm), quantifying the magnitude of the problem. This phenomenon is a serious practical problem for the use of this contact scheme in production. Due to the RTA failure during cooling, the ohmic metal on Batch-2 samples also displayed a significantly rougher morphology and the discolouration due to the oxidation.

(a) Cracks in the ohmic metal stack after ohmic contact RTA anneals on sample C1 in Batch-2

(b) A crater appearing in the places of the crack in the ohmic metal stack after 1ME anneal on sample C1 in Batch-3

Figure 4.8: Ohmic metal morphology problems and their effect on 1ME overlayer on samples C1 in Batch-2 and C1 in Batch-3.

4.5 HEMT DC and Pulsed I-V Curves/ Current Collapse

4.5.1 Batch-1

DC and pulsed I-V transistor curves are shown in Fig. 4.9, 4.10 and 4.11. Gate baseline voltage of -5 V, pulse period of 10 ms and pulse duration of 500 μ s were used. A notable reduction in current collapse on unpassivated samples as well as improved DC current densities were clearly seen on Si-treated samples S1 and S2. Samples O1 and O2 did not demonstrate any improvements in these measurements.

Figure 4.9: Batch-1 DC and pulsed I-V measurements of unpassivated and passivated HEMTs on samples † A1 and B1.

(a) B2 unpassivated

(b) B2 passivated/1ME

(c) S1 unpassivated

(d) S1 passivated/1ME

(e) S2 unpassivated

(f) S2 passivated/1ME

Figure 4.10: Batch-1 DC and pulsed I-V measurements of unpassivated and passivated HEMTs on samples B2, S1 and S2.

(a) O1 unpassivated

(b) O1 passivated/1ME

(c) O2 unpassivated

Figure 4.11: Batch-1 DC and pulsed I-V measurements of unpassivated and passivated HEMTs on samples O1 and O2.

4.5.2 Batch-3

DC and pulsed I-V transistor curves are shown in Fig. 4.12, 4.13 and 4.14. Gate baseline voltage of -8 V (Si-coated samples required lower voltage to fully pinch off), pulse period of 10 ms and pulse duration of 500 μ s were used. The Si-coated samples C1 and C2 displayed a significant reduction in current collapse even before the SiN passivation step. Following passivation, both current densities and current collapse improved even further on samples C1 and C2. Samples S1 and S2 exhibited worse current collapse than seen on control samples before the passivation, in contrast with the results from Batch-1. Additional processing steps performed for the water contact angle measurements were thought to be responsible for this result.

Following the passivation, transistor DC peak current densities for the gate voltage of 2 V reached the value of 915 mA/mm on the control sample, 1,050 mA/mm on Si-treated sample and 1,090 mA/mm on Si-coated sample. These results may be indicative of small 2DEG improvements due to Si treatments and coatings on AlGaIn/GaN material present even after the passivation.

Figure 4.12: Batch-3 DC and pulsed I-V measurements of unpassivated and passivated HEMTs on samples tB1 and tB2 .

Figure 4.13: Batch-3 DC and pulsed I-V measurements of unpassivated and passivated HEMTs on samples S1 and S2

(a) C1 unpassivated

(b) C1 passivated/1ME

(c) C2 unpassivated

(d) C2 passivated/1ME

Figure 4.14: Batch-3 DC and pulsed I-V measurements of unpassivated and passivated HEMTs on samples C1 and C2

4.6 HEMTs Threshold Voltage and Transconductance

4.6.1 Batch-3

Drain current as a function of gate and drain voltages were indicative of both the threshold voltage and the reverse leakage current in HEMTs on Batch-3 samples. The I-V curves are shown in Fig. 4.15 and 4.16. Note that the reverse leakage current is almost independent of drain bias for the Si-treated and Si-coated samples after passivation. Samples [†]B1 and [†]B2 as well as S1 and S2 were in agreement with each other. Leakage currents and threshold voltages were significantly different between samples C1 and C2, which is believed to be due to the additional processing steps performed for the water contact angle measurement on the sample C2. In general there is an increase in the magnitude of the threshold voltage on Si-treated and Si-coated samples. Unlike in the study performed by Onojima *et al.* [69], both the magnitude and changes in threshold voltages are significantly smaller in this study, indicating that there is a significant difference either in the GaN material or transistor structure used by Onojima.

(a) t B1 unpassivated (b) t B1 passivated/1ME

(c) S1 unpassivated

(d) S1 passivated/1ME

Figure 4.15: Batch-3 threshold voltage and reverse bias leakage current measurement on samples t B1 and S1.

(a) C1 unpassivated

(b) C1 passivated/1ME

(c) C2 unpassivated

(d) C2 passivated/1ME

Figure 4.16: Batch-3 threshold voltage and reverse bias leakage current measurement on samples C1 and C2.

Typical plots resulting from HEMT transconductance measurements are shown in Fig. 4.17. Note that these plots were used to extract the actual values of the the threshold (pinch off) voltages of the devices.

Figure 4.17: Batch-3 HEMTs transconductance measurements on samples †B1 and C2.

Table 4.6 summarizes peak transconductances and threshold voltages for Batch-3 HEMT devices. Additional processing steps performed on sample C2 for water contact angle measurement were thought to modify the thin Si film on the AlGaIn surface, resulting in a different threshold voltage.

	After Gate Anneal		Passivation/1ME	
	V_{th} (V)	$g_{m(max)}$ (mS/mm)	V_{th} (V)	$g_{m(max)}$ (mS/mm)
†B1	-3.3	211	-3.4	238
†B2	-2.9	202	-3.0	234
S1	-3.8	205	-4.1	244
S2	-3.3	174	-3.7	238
C1	-4.7	212	-4.9	213
C2	-3.7	239	-3.9	242

Table 4.6: Batch-3 transconductance and threshold voltage summary.

4.7 HEMT Gate Schottky Junction I-V Curves

4.7.1 Batch-1

Fig. 4.18 provides typical I-V curves of gate Schottky junction measurements on Batch-1 samples and Table 4.7 provides the summary of the results. Si-treated samples differed significantly in both ideality factors and saturation current densities under the reverse bias conditions compared to the controls, displaying better ideality factors and higher leakage currents. In addition, reverse currents were effectively flat after -3 V was reached on the gate on samples S1 and S2. This result was consistent

Figure 4.18: Batch-1 gate Schottky junction I-V curves after the passivation and 1ME deposition steps.

with the reduction in Schottky barrier height due to the formation of TiSi_x compounds which in turn increased the thermionic emission current transport component. On the other hand, the anneal of the untreated samples B1 and B2 reduced the leakage

current by a factor of two. The SiON treatment reduced the leakage current by one order of magnitude. Growth of thin oxides on the AlGaN surface is thought to be responsible for this effect.

	After Passivation/1ME		
	$I_{R,-10V}$ (mA/mm)	$I_{F,+3V}$ (mA/mm)	n
†A1	0.45	630	2.30
B1	0.25	550	2.30
B2	0.24	630	2.30
S1	2.4	850	1.26
S2	2.5	775	1.22
O1	0.048	580	1.76

Table 4.7: Batch-1 gate Schottky junction I-V summary after passivation and 1ME dep. processing steps.

4.7.2 Batch-3

Table 4.8 summarizes Batch-3 Schottky junction measurement results. A noticeable difference existed between samples C1 and C2 in leakage currents and ideality factors and is believed to be the result of the additional processing steps performed for the water contact angle measurement on the sample C2.

	After Passivation/1ME			
	$I_{R,-40V}$ (mA/mm)	$I_{R,-10V}$ (mA/mm)	$I_{F,+3V}$ (mA/mm)	n
†B1	3.8	0.44	790	1.95
†B2	3.0	0.85	970	1.95
S1	4.0	2.7	800	1.44
S2	8.6	3.8	840	1.50
C1	46	23	800	1.44
C2	6.7	3.8	610	2.05

Table 4.8: Batch-3 gate Schottky junction I-V summary after passivation and 1ME dep. processing steps.

4.8 Drain-to-Source Breakdown Voltage

4.8.1 Batch-1

The devices on the control and Si- and SiON-treated samples were not found to break down at drain to source voltages of up to 150 V (measurement power limits were reached on some devices before 200 V bias was reached). However, Si-treated samples exhibited higher leakage current than control samples, while SiON-treated samples showed lower leakage than controls for a wide range of the bias conditions. Fig. 4.19 shows typical I-V curves obtained from these measurements.

Figure 4.19: Batch-1 drain-to-source breakdown voltage measurements. Gate voltage (V_g) was set to -5 V.

4.9 SIMS AlGaN/GaN Depth Profiling

The SIMS profile of ^{27}Al on the sample $^{\dagger}\text{B1}$ from Batch-2 was used to establish a rough estimate of the profiling depth as a function of time. The peak in Al signal at the time of 40 seconds corresponded to the AlN spacer under 20 nm of AlGaN barrier layer, giving ion etch rate of 0.5 nm/sec (Fig. 4.20). Fig. 4.21 provides Si SIMS profiles for the reference sample R1, control sample $^{\dagger}\text{B1}$ and Si-treated samples S1, L1 and H1 after the ohmic contact deposition and ohmic annealing processing steps. The difference in profiles between control sample $^{\dagger}\text{B1}$ (which had no Si surface treatments) and the other samples is striking. It is clear that even on samples S1, L1 and H1, which were subjected to an aggressive Si wet etch, very large amounts of Si remain on the surface region. For all these samples the initial Si signal is comparable to that of reference R1, which had pure Si left in place on the surface.

The thickness of the remnant Si rich film scaled according to RTA temperatures. Higher temperatures produced thicker remnant films as can be seen on the first 5 nm of the Si SIMS profile. Note that the profiling depth inferred from the SIMS profile of Al was only a rough approximation in Si profiles due to different etching rates of AlGaN and Si films. Although the target thickness for the Si film on the reference sample was 5 nm, surface profiling with Dektak resulted in a value of 7.3 nm. Thicker than 5 nm film thickness values were also obtained on Batch-3 samples, likely resulting from the growth of native oxide on the Si surface. By comparing the profiles of the Si-treated samples to the reference sample, S1 and H1 were estimated to contain about 2.4 nm of the Si-containing film.

All Si-treated samples showed a long “tail” of Si signal apparently extending to a depth of almost 20 nm. This is almost certainly an artifact resulting from forward scattering of Si from the surface layer, driven by the Cs ion analysis beam. Note that samples L1 and H1 show very similar Si profiles, despite the large difference in

anneal temperatures.

Figure 4.20: Batch-2 SIMS ^{27}Al profile of the control sample $^\dagger\text{B1}$.

Figure 4.21: Batch-2 SIMS ^{28}Si profile of the control sample as well as Si-treated and Si-coated reference samples.

4.10 AlGaN Surface Chemical Composition and Film Thickness by XPS and AR-XPS

Shirley background subtraction was performed on all XPS spectra. The elemental Si $2p$ peak at 99.3 eV was used for charge referencing. The spin-orbit splitting ratio of 1:2 with the peaks separated by 0.6 eV was used for fitting of the Si $2p$ peaks. On samples which did not contain elemental Si, Ga $3p$ peaks with the area ratios of 1:2 due to the spin-orbit splitting of the p orbital in the binding energy window of 105 eV to 110 eV were used for spectra alignment. The position of the Ga $3p$ peaks with respect to the Si $2p$ peak was established from the Ar sputtered reference sample R2, Batch-2.

Additional peaks were used where necessary to fit the XPS data in order to account for the Si^{1+} , Si^{2+} and Si^{3+} components [76], where the Si in elemental state and in SiO_2 compound possessed Si^0 and Si^{4+} oxidation states respectively.

The Gaussian-Lorentzian peak shape fitting was initially used, however, the best fits were achieved when 100% Gaussian peak shape fitting was performed.

Fig. 4.22 provides the XPS spectra and peak fitting results for sample R2 at three stages of Ar sputtering. By setting the peak location to 99.3 eV for elemental Si, Ga $3p$ peaks were found to be located at 108.9 and 105.6 eV. Note that the two Si $2p$ peaks for SiO_x compounds were not sufficient to accurately represent all oxidation states of Si present in Ar sputtered film. However, the analysis relied on the elemental Si peak, the location of which was not significantly affected by the fitting of SiO_x peaks. The elemental Si peak had a constant Full Width at Half Maximum (FWHM).

(a) Before Ar sputtering

Peak	Position eV	Area	FWHM eV
Si2p(SiO ₂)	103.2	989.9	1.40
Si2p(SiO _x)	101.2	527.3	3.08
Si2p(elem)	99.3	1175.4	0.85

(b) Before Ar sputtering peak parameters

(c) After 10 min of Ar sputtering

Peak	Position eV	Area	FWHM eV
Si2p(SiO _x I)	102.4	686.7	1.60
Si2p(SiO _x II)	100.9	867.4	1.84
Si2p(elem)	99.3	1307.6	0.85

(d) After 10 min of Ar sputtering peak parameters

(e) After 30 min of Ar sputtering

Peak	Position eV	Area	FWHM eV
Ga3p 1/2(AlGaIn)	108.9	579.2	2.49
Ga3p 3/2(AlGaIn)	105.6	1158.4	2.49
Si2p(SiO _x I)	102.5	449.7	1.60
Si2p(SiO _x II)	100.9	951.6	2.15
Si2p(elem)	99.3	807.2	0.85

(f) After 30 min of Ar sputtering peak parameters

Figure 4.22: Batch-2 R2 XPS spectra before, after 10 and 30 min of Ar sputtering.

A Si 2p peak was observed on Si-treated and wet etched samples that did not belong to either SiO₂ or elemental Si (Fig. 4.23(a)). Because of the aggressiveness of HF:HNO₃ wet etch and based on the location of the peak [77] (Fig. 4.24), it was concluded that this peak was associated with a non-stoichiometric SiN_x containing surface film. Fig. 4.24 Additionally, the ohmic contact processing steps resulted in an additional peak corresponding to SiO₂ (103.4 eV), indicating that the film was easily oxidized (Fig. 4.23(c)).

(a) Remnant SiN_x film on the sample S2 after wet etch

Peak	Position eV	Area	FWHM eV
Ga3p 1/2(AlGaIn)	109.1	1005.3	2.49
Ga3p 3/2(AlGaIn)	105.6	2010.5	2.49
Si2p(SiN _x)	102.2	1560.8	2.12

(b) Remnant SiN_x peak parameters

(c) Remnant SiN_x/SiO₂ film on the sample S2 after processing steps leading to ohmic contact metal deposition

Peak	Position eV	Area	FWHM eV
Ga3p 1/2(AlGaIn)	109.0	655.4	2.49
Ga3p 3/2(AlGaIn)	105.6	1310.7	2.49
Si2p(SiO ₂)	103.4	901.9	1.41
Si2p(SiN _x)	102.2	493.2	1.77

(d) Remnant SiN_x/SiO₂ peak parameters

Figure 4.23: Batch-3 S2 XPS spectra before and after processing steps leading to ohmic contact metal deposition.

Figure 4.24: The location of several Si peaks from peak fitting by Matsuoka *et al.* [77].

The AR-XPS surface composition measurement was performed on sample S2 from Batch-3 after the water contact angle measurement. The spectra recorded for take off angles of 30, 45 and 77.5 degrees are shown in Fig. 4.25. Assuming that most of the SiN_x was consumed and using a photoelectric attenuation length of 3 nm for SiO_2 film [51], $\beta_{\text{Si}/\text{Ga}}$ for 2*p* and 3*p* peaks of $0.27/0.84 = 0.32$ [78], a fit between the AR-XPS equation (Eq. 2.45) and the experimental data was obtained (Fig. 4.26). The best fit resulted in a film thickness of 2.45 nm, in agreement with SIMS results.

An additional way to estimate the thickness of the oxidized SiN_x film is to compare its peak intensity to the Si peak intensity of the native SiO_2 film on the sample R2 from Batch-2. The combined areas of Si peaks produced by remnant film (SiN_x and SiO_2) and native SiO_2 were (902+493) and 990 respectively. Given a native oxide thickness of 1.5 - 2.0 nm, the estimate of the remnant SiN_x film thickness was 2.1 - 2.8 nm. The oxidation had likely increased the overall thickness of the film, thus the original SiN_x film was thinner.

Film thickness obtained from SIMS and AR-XPS measurements as well as from the comparison to the native oxide peak agreed. A rough final estimate of the initial (non-oxidized) SiN_x film thickness was 1.0 - 2.0 nm.

(a) Spectra at 30° take off angle

Peak	Position eV	Area	FWHM eV
Ga3p(AlGa _N)	108.9	254.2	2.49
Ga3p(AlGa _N)	105.5	508.4	2.49
Si2p(SiO ₂)	103.3	657.7	1.41
Si2p(SiN _x)	102.2	340.7	1.77

(b) 30° peak parameters

(c) Spectra at 45° take off angle

Peak	Position eV	Area	FWHM eV
Ga3p(AlGa _N)	108.9	655.4	2.49
Ga3p(AlGa _N)	105.5	1310.7	2.49
Si2p(SiO ₂)	103.3	905.0	1.41
Si2p(SiN _x)	102.2	490.2	1.77

(d) 45° peak parameters

(e) Spectra at 77.5° take off angle

Peak	Position eV	Area	FWHM eV
Ga3p(AlGa _N)	109.0	1881.4	2.49
Ga3p(AlGa _N)	105.6	3762.8	2.49
Si2p(SiO ₂)	103.3	1519.8	1.41
Si2p(SiN _x)	102.2	813.3	1.77

(f) 77.5° peak parameters

Figure 4.25: Batch-3 S2 AR-XPS spectra at 30°, 45° and 77.5° take off angle.

Figure 4.26: The fit of the AR-XPS equation to the experimental data using Si to Ga sensitivity ratio parameter β of 0.32, photoelectron attenuation length λ of 3 nm and the film thickness d of 2.45 nm. Theoretical curves with the values of $d = 1.95$ nm and $d = 2.95$ nm are also shown for comparison.

The combined plot of XPS spectra for several samples from Batch-2 and Batch-3 is shown in Fig. 4.27. The oxidation of the remnant Si film at several stages of processing can be clearly seen.

Figure 4.27: Combined plot of XPS spectra for several samples from Batch-2 and Batch-3

4.11 AlGaN Surface Water Contact Angle

The similarity between Si-coated and Si-treated samples is clearly seen from the water contact angle measurement results (Fig. 4.28). Hydrophilic surfaces are characteristic to oxides and Si-treated and coated samples were expected to grow oxides of Si in the top most layers. This measurement clearly distinguishes the control samples from the Si-treated or coated samples, which further signified the presence of the remnant Si-containing film on the Si-treated samples S1 and S2.

The difference in contact angles between the control samples and Si-treated/coated samples at each of the last three processing steps have very significant implications. Firstly, there was already a contact angle difference of 15° following the “pre ohmic cleaning”. Thus the actual clean could be affected by the hydrophilic properties of the surface at that step. Moreover, following Deionized (DI) water rinse, the contact angle difference exceeded 20° . The additional 20 hours of storage time resulted in the contact angle of about 45° on control samples while the angle remained at about 5° on Si-treated and coated samples. The time period between “pre ohmic clean” and ohmic metal evaporation can be crucial for control samples, affecting the reproducibility of the ohmic contact resistance values. Si-treated and Si-coated AlGaN surfaces might be insensitive to the time between these steps, based on the fairly constant water contact angle value displayed by these samples.

Figure 4.28: Batch-3 water contact angle measurement result.

Chapter 5

DISCUSSION

The rigorous investigation of the AlGaIn surface treatments and coatings with Si films revealed several important findings which were not previously reported in the literature. The results obtained from SIMS profiling, XPS surface composition and contact angle measurements on the samples following Si film deposition, RTA annealing and film removal, clearly indicated that a remnant Si-containing film was left on the surface. Given the aggressiveness of the HF:HNO₃ wet etch towards both Si and SiO₂, elemental or oxidized Si were ruled out as the candidates for the composition of the remnant film. The location of the Si 2*p* peak on the Si-treated samples in XPS spectrum further confirmed that the Si in the film did not belong to either one of those states but belonged to a non-stoichiometric SiN_{*x*} compound. The formation of this SiN_{*x*} compound under high temperature annealing is easily justified when the enthalpy of formations are examined. The enthalpy of formation of Si₃N₄ is -850 kJ/mol [79] and TiN is -265 kJ/mol. Since there is overwhelming proof of Ti forming TiN compounds during ohmic contact annealing, Si₃N₄ should be easily formed given its lower enthalpy of formation, especially considering high annealing temperatures used in this study. The film was found to be easily oxidized upon further processing. The low stability of the film was attributed to its non-stoichiometric structure and enthalpy of formation for SiO₂, -911 kJ/mol, lower than for Si₃N₄. The

thickness of the remnant film (prior to any oxidation) was estimated to be between 1.0 and 2.0 nm from both XPS and SIMS measurement results.

The inclusion of the Si-coated reference sample clearly revealed the SIMS measurement artifact on Si-treated samples. Without the reference, the shape of the Si profile in the SIMS measurements might be attributed to the thermal diffusion of Si. However, by comparing the profiles of Si-treated samples to the reference sample it can be seen that the solid state diffusion, if any, was heavily masked by the measurement artifact arising from the interaction of a primary ion beam and the remnant Si-containing film.

The discovery of this remnant film resulting from the reaction between Si and AlGaIn layers opens new avenues to understanding how the annealing of the Si thin film on AlGaIn surfaces, which was thought to result in solid state diffusion, can improve or alter electrical performance of fabricated devices. It is known that Si acts as a bulk n-type dopant in the AlGaIn layer. It has also been suggested that Si acts as a donor on the AlGaIn surface providing passivating results similar to those achieved with Si_3N_4 [69, 71]. Since the devices on Si-coated samples produced similar, if not better, results in the current collapse and current density measurements, diffusion of Si into the epilayer as a dominant mechanism for electrical enhancements appears unlikely. The evidence gathered here indicates that Si acts as a shallow surface sheet donor, providing the electrons to the 2DEG and/or to surface electron traps. This in turn reduced the current collapse and improved the 2DEG densities. However, the actual mechanism by which Si acts as a surface donor is still not clear.

In a fashion similar to the formation of the tunnelling ohmic contacts in which the reaction of titanium and nitrogen creates an AlGaIn surface heavily n-type doped with nitrogen vacancies, the reaction between Si and AlGaIn, resulting in SiN_x formation, might also produce a surface heavily n-type doped with nitrogen vacancies. Given the temperatures at which TiN forms during ohmic contact metal alloying, the reaction

between Si and AlGaIn should readily progress at temperatures in the range of 800 to 1,000°C. Based on enthalpy of formation, Si and AlGaIn materials should react more strongly than Ti and AlGaIn, resulting in a heavier doping of the AlGaIn surface.

The water contact angle measurement has revealed the similarity between Si-coated and Si-treated samples, and the results were in sharp contrast to the measurements on the control samples. Combined with the XPS and SIMS measurements, this finding is also explained by the remnant $\text{SiN}_x/\text{SiO}_x$ film left on the AlGaIn surface. Of particular interest are the results after the last few processing steps leading to the ohmic metal deposition. There is a clear change in the contact angle on the control samples. The water contact angle remains fairly constant on Si-treated and coated samples after the pre ohmic metal clean, followed by the DI rinse and 20 hours of storage time, in contrast to the contact angle on the control samples. This result has two implications. First, reproducibility is much easier achieved on Si-treated and coated samples given that the hydrophilic properties of their surfaces do not change significantly with time. Secondly, stronger surface hydrophilic properties can enhance the cleaning of the surface resulting in better removal of common contaminants.

The ohmic contact resistances were not as sensitive to high alloying temperatures on the Si-treated and coated samples. The remnant Si may thus require slightly higher temperatures to become an active part of the complete ohmic contact or it may slow down the rate of certain reactions occurring during high annealing temperatures. Si-treated and coated samples demonstrated far superior characteristics than the control sample upon heavy oxidation of the ohmic metal in Batch-2. These results indicate that Si treatments and coatings can improve the ohmic contacts and device current densities in unfavourable processing conditions for the ohmic contact metal. Given that NRC's processing for ohmic contacts is very well optimized for the material used in this study, it may explain why other groups witnessed substantial improvements to their ohmic contacts.

Under proper processing conditions similar ohmic contact resistances were obtained on both Si-treated, coated and control samples, indicating that the remnant film was not providing any significant improvements to the ohmic contacts. The lowest ohmic contact resistances achieved in this study are in the range of 0.5 to 0.6 Ohm-mm. Assuming the 2DEG sheet resistance under the ohmic contact (R_{sc}) is equal to the 2DEG sheet resistance on unpassivated material R_s (~ 400 Ohm/sq), a specific contact resistance in the range of 6.3×10^{-6} to 9.0×10^{-6} Ohm-cm² was obtained using Eq. 2.39 and Eq. 2.41.

The passivating properties of the Si treatment were seen in the Hall measurements as well as in the 2DEG sheet resistance measurements on TLM and HEMT devices. Si-treated and Si-coated surfaces were found to be pre-passivated before the SiN passivation step, which is beneficial for in-process device performance monitoring and adjustments. The HEMTs exhibited noticeably lower current collapse and higher saturation current densities on both Si-treated (Batch-1) and Si-coated (Batch-3) samples before the passivation. In addition, there was evidence that Si treatments and coatings improve the 2DEG densities even after the SiN passivation. The 2DEG sheet resistances were noticeably better on Si-treated samples in Batch-1 and HEMTs had higher saturation current densities on both Si-treated and Si-coated samples from Batch-1 and Batch-3. The peak saturation current densities for $V_g = 2$ V reached values of 915 mA/mm on the passivated control sample, 1,050 mA/mm on the Si-treated sample and 1,090 mA/mm on the Si-coated sample in Batch-3. These results are in agreement with findings by Onojima *et al.* [69], where Si films were used for passivation. Similar current enhancements to the current densities were obtained: from 1,100 mA/mm to 1,200 mA/mm.

Several electrical measurements revealed high gate and drain leakage currents on Si-treated and coated samples. Gate Schottky junction ideality factors were significantly lower for Si-treated samples, indicating that the thermionic emission became

a more dominant mechanism of current conduction. This result is further supported by higher reverse bias leakage currents obtained on Si-treated samples. In addition, the shape of the reverse bias I-V curve for Si-treated samples were significantly flatter, indicating a more ideal Schottky junction behaviour conforming better to the ideal diode equation. Higher leakages and the dominant thermionic current transport mechanism are consistent with the lowering of the Schottky barrier height, as can be seen in Eq. 2.28. Higher leakage currents on Si-treated and Si-coated samples can also be explained by an additional leakage path on top of the AlGaIn epilayer. This leakages may have lead to filling of surface states that passivate the surface thereby increasing 2DEG densities in the channel. The magnitude of the leakages was in the range of a few mA/mm, not sufficient to account for the improvements in 2DEG sheet resistance values or HEMT peak drain current densities following the AlGaIn surface treatment or coating with Si.

There was a distinct change in threshold voltage in fabricated devices on Si-treated and Si-coated samples. This shift, which is also observed in metal-insulator-semiconductor HEMT devices where an insulator is placed under the gate, can be explained by the presence of remnant SiN_x and Si films on the AlGaIn surface.

The experiments did not reveal any significant improvements to AlGaIn/GaN device performance following the surface treatment with SiON films. Therefore it was concluded that the SiON films were intentionally left in place for further processing in the previous studies. The benefits of using SiON instead of Si could come from the ease with which the film can be wet etched in comparison to Si or Si_3N_4 . However, the leakage currents were significantly reduced on SiON-treated samples as well as otherwise untreated RTA annealed samples. The formation of very thin oxide films on AlGaIn surface is thought to be responsible for these results.

A few disadvantages of Si treatments and coatings of AlGaIn surfaces are longer

process flows, the requirement of better quality control of Si deposition (as demonstrated by the cracking of the ohmic metal in Batch-2 and Batch-3 samples) and higher leakage currents in fabricated devices.

Chapter 6

CONCLUSIONS AND FUTURE WORK

6.1 Conclusion

In conclusion, the effects of Si treatments and coatings on AlGaIn surfaces were rigorously studied. Several important findings were made and presented. Si treatment resulted in a remnant 1.0 - 2.0 nm thick SiN_x film left on the AlGaIn surface. SIMS measurement artifact due to this film was demonstrated. The water contact angle measurement provided an additional evidence for the difference between the Si-treated, Si-coated and control samples. In addition, the contact angles varied substantially on the control samples in sharp contrast to the contact angle on the treated and coated surfaces. The electrical changes in fabricated devices were attributed to the remnant or intentionally left Si film, not the solid state diffusion. The electrical measurement results demonstrated pre-passivating properties of Si treatments and coatings. In addition, Si treatment and coatings produced better ohmic contact resistances in adverse processing conditions. Under proper processing conditions similar ohmic contact resistance values of about 0.5 - 0.6 Ohm-mm were obtained on treated, coated and control samples. Si-treated and coated samples displayed higher leakage currents and more negative threshold voltages.

6.2 Future Work

This study did not reveal how exactly Si acted as a surface sheet donor. The parameters such as a donor energy level and density may be established by measuring the 2DEG channel resistance and carrier density versus temperature in a cryostat. Additionally, performing the Si treatments and coatings using thinner AlGaIn epilayers might result in more drastic enhancements, revealing changes to the device performance undetected in this study.

Another area for future work lies in the determination of Schottky barrier heights and ideality factors in Si-treated and Si-coated samples. Although this study suggests that the Schottky barrier height is reduced on Si-treated and Si-coated samples, the evidence is neither overwhelming nor complete.

In order to successfully use the Si treatments and coatings in production, the problem with cracking of the ohmic metal stack has to be resolved. Because this effect was not noticed during the first fabrication run, it might be possible to fully eliminate it. One possible way to address this problem is to evaporate Si *in-situ* prior to ohmic contact deposition.

List of References

- [1] U. K. Mishra, L. Shen, T. E. Kazior, and Y. Wu, “GaN-based RF power devices and amplifiers,” *Proceedings of the IEEE*, vol. 96, no. 2, pp. 287–305, 2008.
- [2] O. Ambacher, “Growth and applications of group III-nitrides,” *Journal of Physics D: Applied Physics*, vol. 31, no. 20, pp. 2653–2710, 1998.
- [3] T. Nomura, M. Masuda, N. Ikeda, and S. Yoshida, “Switching characteristics of GaN HFETs in a half bridge package for high temperature applications,” *IEEE Transactions on Power Electronics*, vol. 23, no. 2, pp. 692–697, 2008.
- [4] S. J. Pearton, C. R. Abernathy, and F. Ren, *Gallium Nitride Processing for Electronics, Sensors and Spintronics*. London, England: Springer-Verlag, 2006.
- [5] R. Quay, *Gallium Nitride Electronics*. Berlin, Germany: Springer, 2008.
- [6] P. Jonnard, N. Capron, F. Semon, J. Massies, E. Martinez-Guerrero, and H. Mariette, “Electronic structure of wurtzite and zinc-blende AlN,” *The European Physical Journal B*, vol. 42, pp. 351–359, 2004.
- [7] “V semiconductors optics and optoelectronics I.” <http://kottan-labs.bgsu.edu/teaching/workshop2001/chapter5.htm>, 2001.
- [8] S. C. Jain, M. Willander, J. Narayan, and R. Van Overstraeten, “III-nitrides: Growth, characterization, and properties,” *Journal of Applied Physics*, vol. 87, no. 3, pp. 965–1006, 2000.
- [9] H. T. Wang, L. S. Tan, and E. F. Chor, “Optical and electrical characterization of annealed silicon-implanted GaN,” *Semiconductor Science and Technology*, vol. 19, no. 2, pp. 142–146, 2004.
- [10] J. D. Plummer, M. D. Deal, and P. B. Griffin, *Silicon VLSI Technology*. Upper Saddle River, NJ USA: Prentice Hall Inc, 2000.

- [11] X. A. Cao, S. J. Pearton, R. K. Singh, C. R. Abernathy, J. Han, R. J. Shul, D. J. Rieger, J. C. Zolper, R. G. Wilson, M. Fu, J. A. Sekhar, H. J. Guo, and S. J. Pennycook, "Rapid thermal processing of implanted GaN up to 1500 °C," *Materials Research Society Symposium - Proceedings*, vol. 537, 1999.
- [12] Y. Mishin and C. Herzig, "Grain boundary diffusion: Recent progress and future research," *Materials Science and Engineering A*, vol. 260, no. 1-2, pp. 55–71, 1999.
- [13] R. Jakiela, A. Barcz, E. Dumiszewska, and A. Jagoda, "Si diffusion in epitaxial GaN," *Physica Status Solidi (C) Current Topics in Solid State Physics*, vol. 3, pp. 1416–1419, 2006.
- [14] P. H. Ladbrooke, *MMIC Design: GaAs FETs and HEMTs*. Norwood, MA USA: Artech house Inc, 1989.
- [15] S. M. Sze, *High-Speed Semiconductor Devices*. USA: John Wiley & Sons Inc, 1990.
- [16] M. Asif Khan, J. N. Kuznia, A. R. Bhattarai, and D. T. Olson, "Metal semiconductor field effect transistor based on single crystal GaN," *Applied Physics Letters*, vol. 62, no. 15, pp. 1786–1787, 1993.
- [17] M. Asif Khan, A. Bhattarai, J. N. Kuznia, and D. T. Olson, "High electron mobility transistor based on a GaN-AlGa_N heterojunction," *Applied Physics Letters*, vol. 63, no. 9, pp. 1214–1215, 1993.
- [18] J. A. Bardwell, T. Lester, H. Tang, B. Surridge, A. Harrison, and A. Lau, "Al-GaN/GaN transistors and circuits." Power Point Presentation, 2009.
- [19] F. Bernardini, V. Fiorentini, and D. Vanderbilt, "Spontaneous polarization and piezoelectric constants of III-V nitrides," *Physical Review B - Condensed Matter and Materials Physics*, vol. 56, no. 16, 1997.
- [20] O. Ambacher, B. Foutz, J. Smart, J. R. Shealy, N. G. Weimann, K. Chu, M. Murphy, A. J. Sierakowski, W. J. Schaff, L. F. Eastman, R. Dimitrov, A. Mitchell, and M. Stutzmann, "Two dimensional electron gases induced by spontaneous and piezoelectric polarization in undoped and doped AlGa_N/Ga_N heterostructures," *Journal of Applied Physics*, vol. 87, no. 1, pp. 334–344, 2000.
- [21] O. Ambacher, J. Majewski, C. Miskys, A. Link, M. Hermann, M. Eickhoff, M. Stutzmann, F. Bernardini, V. Fiorentini, V. Tilak, B. Schaff, and L. F.

- Eastman, "Pyroelectric properties of Al(In)GaN/GaN hetero- and quantum well structures," *Journal of Physics Condensed Matter*, vol. 14, no. 13, pp. 3399–3434, 2002.
- [22] N. Li, D. G. Zhao, and H. Yang, "Polarization effects simulation of AlGa_N/Ga_N heterojunction by using a symbolistic δ -doping layer," *Solid State Communications*, vol. 132, no. 10, pp. 701–705, 2004.
- [23] I. P. Smorchkova, C. R. Elsass, J. P. Ibbetson, R. Vetury, B. Heying, P. Fini, E. Haus, S. P. DenBaars, J. S. Speck, and U. K. Mishra, "Polarization-induced charge and electron mobility in AlGa_N/Ga_N heterostructures grown by plasma-assisted molecular-beam epitaxy," *Journal of Applied Physics*, vol. 86, no. 8, pp. 4520–4526, 1999.
- [24] J. P. Ibbetson, P. T. Fini, K. D. Ness, S. P. DenBaars, J. S. Speck, and U. K. Mishra, "Polarization effects, surface states, and the source of electrons in AlGa_N/Ga_N heterostructure field effect transistors," *Applied Physics Letters*, vol. 77, no. 2, pp. 250–252, 2000.
- [25] L. Shen, S. Heikman, B. Moran, R. Coffie, N. . Zhang, D. Buttari, I. P. Smorchkova, S. Keller, S. P. DenBaars, and U. K. Mishra, "Algan/aln/gan high-power microwave hemt," *IEEE Electron Device Letters*, vol. 22, no. 10, pp. 457–459, 2001.
- [26] T. . Yu and K. F. Brennan, "Monte carlo calculation of two-dimensional electron dynamics in Ga_N-AlGa_N heterostructures," *Journal of Applied Physics*, vol. 91, no. 6, pp. 3730–3736, 2002.
- [27] C. S. Chang and H. R. Fetterman, "An analytic model for high-electron-mobility transistors," *Solid State Electronics*, vol. 30, no. 5, pp. 485–491, 1987.
- [28] F. Roccaforte, F. Giannazzo, F. Iucolano, J. Eriksson, M. H. Weng, and V. Raineri, "Surface and interface issues in wide band gap semiconductor electronics," *Applied Surface Science*, vol. 256, no. 19, pp. 5727–5735, 2010.
- [29] N. Chaturvedi, U. Zeimer, J. Wrfl, and G. Trnkle, "Mechanism of ohmic contact formation in AlGa_N/Ga_N high electron mobility transistors," *Semiconductor Science and Technology*, vol. 21, no. 2, pp. 175–179, 2006.
- [30] S. O. Kasap, *Principles of Electronic Materials and Devices*. USA: McGraw-Hill, 2006.

- [31] M. Shur, S. Rumyantsev, and M. Levinshtein, *SiC materials and devices*. USA: World Scientific Publishing, 2006.
- [32] D. Qiao, L. S. Yu, S. S. Lau, J. M. Redwing, J. Y. Lin, and H. X. Jiang, "Dependence of Ni/AlGaN schottky barrier height on Al mole fraction," *Journal of Applied Physics*, vol. 87, no. 2, pp. 801–804, 2000.
- [33] E. H. Rhoderick, "Metal-semiconductor contacts," *IEE Proceedings I: Solid State and Electron Devices*, vol. 129, no. 1, pp. 1–14, 1982.
- [34] Z. Fan, S. N. Mohammad, W. Kim, A. Aktas, A. E. Botchkarev, K. Suzue, H. Morko, K. Duxstad, and E. E. Haller, "Ohmic contacts and Schottky barriers to n-GaN," *Journal of Electronic Materials*, vol. 25, no. 11, pp. 1703–1708, 1996.
- [35] S. N. Mohammad, "Contact mechanisms and design principles for nonalloyed ohmic contacts to n-GaN," *Journal of Applied Physics*, vol. 95, no. 9, pp. 4856–4865, 2004.
- [36] N. Stavitski, M. J. H. Van Dal, R. A. M. Wolters, A. Y. Kovalgin, and J. Schmitz, "Specific contact resistance measurements of metal-semiconductor junctions," vol. 2006 of *International Conference on Microelectronic Test Structures*, pp. 13–17, 2006.
- [37] G. S. Marlow and M. B. Das, "The effects of contact size and non-zero metal resistance on the determination of specific contact resistance," *Solid State Electronics*, vol. 25, no. 2, pp. 91–94, 1982.
- [38] S. Ruvimov, Z. Liliental-Weber, J. Washburn, K. J. Duxstad, E. E. Haller, Z. Fan, S. N. Mohammad, W. Kim, A. E. Botchkarev, and H. Morko, "Microstructure of Ti/Al and Ti/Al/Ni/Au ohmic contacts for n-GaN," *Applied Physics Letters*, vol. 69, no. 11, pp. 1556–1558, 1996.
- [39] A. Motayed, R. Bathe, M. C. Wood, O. S. Diouf, R. D. Vispute, and S. Noor Mohammad, "Electrical, thermal, and microstructural characteristics of Ti/Al/Ti/Au multilayer ohmic contacts to n-type GaN," *Journal of Applied Physics*, vol. 93, no. 2, pp. 1087–1094, 2003.
- [40] B. Van Daele, G. Van Tendeloo, W. Ruythooren, J. Derluyn, M. R. Leys, and M. Germain, "The role of Al on ohmic contact formation on n-type GaN and AlGaN," *Applied Physics Letters*, vol. 87, no. 6, 2005.

- [41] B. P. Luther, S. E. Mohny, T. N. Jackson, M. Asif Khan, Q. Chen, and J. W. Yang, "Investigation of the mechanism for ohmic contact formation in al and ti/al contacts to n-type gan," *Applied Physics Letters*, vol. 70, no. 1, pp. 57–59, 1997.
- [42] J. A. Bardwell, G. I. Sproule, Y. Liu, H. Tang, J. B. Webb, J. Fraser, and P. Marshall, "Comparison of two different Ti/Al/Ti/Au ohmic metallization schemes for AlGa_N/Ga_N," *Journal of Vacuum Science and Technology B: Microelectronics and Nanometer Structures*, vol. 20, no. 4, pp. 1444–1447, 2002.
- [43] N. Miura, T. Oishi, T. Nanjo, M. Suita, Y. Abe, T. Ozeki, H. Ishikawa, and T. Egawa, "Effects of interfacial thin metal layer for high-performance Pt-Au-based Schottky contacts to AlGa_N-Ga_N," *IEEE Transactions on Electron Devices*, vol. 51, no. 3, pp. 297–303, 2004.
- [44] N. Miura, T. Nanjo, M. Suita, T. Oishi, Y. Abe, T. Ozeki, H. Ishikawa, T. Egawa, and T. Jimbo, "Thermal annealing effects on Ni/Au based schottky contacts on n-Ga_N and AlGa_N/Ga_N with insertion of high work function metal," *Solid-State Electronics*, vol. 48, no. 5, pp. 689–695, 2004.
- [45] H. Hasegawa, T. Inagaki, S. Ootomo, and T. Hashizume, "Mechanisms of current collapse and gate leakage currents in AlGa_N/Ga_N heterostructure field effect transistors," *Journal of Vacuum Science and Technology B: Microelectronics and Nanometer Structures*, vol. 21, no. 4, pp. 1844–1855, 2003.
- [46] S. Haffouz, H. Tang, J. A. Bardwell, E. M. Hsu, J. B. Webb, and S. Rolfe, "Al-Ga_N/Ga_N field effect transistors with C-doped Ga_N buffer layer as an electrical isolation template grown by molecular beam epitaxy," *Solid-State Electronics*, vol. 49, no. 5, pp. 802–807, 2005.
- [47] F. Gonzlez-Posada, J. A. Bardwell, S. Moisa, S. Haffouz, H. Tang, A. F. Braa, and E. Muoz, "Surface cleaning and preparation in AlGa_N/Ga_N-based HEMT processing as assessed by X-ray photoelectron spectroscopy," *Applied Surface Science*, vol. 253, no. 14, pp. 6185–6190, 2007.
- [48] I. Cimalla, F. Will, K. Tonisch, M. Niebelschitz, V. Cimalla, V. Lebedev, G. Kitzler, M. Himmerlich, S. Krischok, J. A. Schaefer, M. Gebinoga, A. Schober, T. Friedrich, and O. Ambacher, "AlGa_N/Ga_N biosensor-effect of device processing steps on the surface properties and biocompatibility," *Sensors and Actuators, B: Chemical*, vol. 123, no. 2, pp. 740–748, 2007.

- [49] K. J. Kim, J. S. Jang, D. W. Moon, and H. J. Kang, "A method to determine the interface position and layer thickness in SIMS depth profiling of multilayer films," *Metrologia*, vol. 47, no. 3, pp. 253–261, 2010.
- [50] J. F. Moulder, W. F. Stickle, P. E. Sobol, and K. D. Bomben, *Handbook of X-ray Photoelectron Spectroscopy*. USA: Perkin-Elmer Corporation, 1992.
- [51] Z. H. Lu, J. P. McCaffrey, B. Brar, G. D. Wilk, R. M. Wallace, L. C. Feldman, and S. P. Tay, "SiO₂ film thickness metrology by x-ray photoelectron spectroscopy," *Applied Physics Letters*, vol. 71, pp. 2764–2766, 1997.
- [52] H.-J. Butt, K. Graf, and M. Kappl, *Physics and chemistry of interfaces*. USA: Wiley-VCH, 2003.
- [53] C. Buchheim, V. Lebedev, V. Yanev, G. Kittler, M. Fischer, G. Ecke, O. Ambacher, V. Cimalla, S. Krischok, and J. A. Schaefer, "Surface modifications of AlGa_N/Ga_N sensors for water base nano- and picodroplets," vol. 2, pp. 1007–1010, 2004.
- [54] R. G. Wilson, S. J. Pearton, C. R. Abernathy, and J. M. Zavada, "Thermal stability of implanted dopants in Ga_N," *Applied Physics Letters*, p. 2238, 1995.
- [55] C. Nguyen, P. Shah, E. Leong, M. Derenge, and K. Jones, "Si implant-assisted ohmic contacts to Ga_N," *Solid-State Electronics*, vol. 54, no. 10, pp. 1227–1231, 2010.
- [56] J. Burm, K. Chu, W. A. Davis, W. J. Schaff, L. F. Eastman, and T. J. Eustis, "Ultra-low resistive ohmic contacts on n-Ga_N using Si implantation," *Applied Physics Letters*, vol. 70, no. 4, pp. 464–466, 1997.
- [57] F. Recht, L. McCarthy, S. Rajan, A. Chakraborty, C. Poblenz, A. Corrion, J. S. Speck, and U. K. Mishra, "Nonalloyed ohmic contacts in AlGa_N/Ga_N HEMTs by ion implantation with reduced activation annealing temperature," *IEEE Electron Device Letters*, vol. 27, no. 4, pp. 205–207, 2006.
- [58] S. J. Pearton, H. Cho, J. R. LaRoche, F. Ren, R. G. Wilson, and J. W. Lee, "Oxygen diffusion into SiO₂-capped Ga_N during annealing," *Applied Physics Letters*, vol. 75, no. 19, pp. 2939–2941, 1999.
- [59] C. F. Lin, H. C. Cheng, G. C. Chi, C. J. Bu, and M. S. Feng, "Improved contact performance of Ga_N film using si diffusion," *Applied Physics Letters*, vol. 76, no. 14, pp. 1878–1880, 2000.

- [60] H. Ishida, Y. Hirose, T. Murata, Y. Ikeda, T. Matsuno, K. Inoue, Y. Uemoto, T. Tanaka, T. Egawa, and D. Ueda, "A high-power RF switch IC using AlGa_N/Ga_N HFETs with single-stage configuration," *IEEE Transactions on Electron Devices*, vol. 52, no. 8, pp. 1893–1899, 2005.
- [61] T. Murata, M. Hikita, H. Ishida, Y. Hirose, K. Inoue, Y. Uemoto, T. Tanaka, and D. Ueda, "Reduction of ohmic contact resistance of AlGa_N/Ga_N HFETs by doping of thermally diffused Si," International Meeting for Future of Electron Devices, p. 47, 2004.
- [62] C. J. Pan, G. C. Chi, B. J. Pong, J. K. Sheu, and J. Y. Chen, "Si diffusion in p-Ga_N," *Journal of Vacuum Science and Technology B: Microelectronics and Nanometer Structures*, vol. 22, no. 4, pp. 1727–1730, 2004.
- [63] S. Jang, F. Ren, S. J. Pearton, B. P. Gila, M. Hlad, C. R. Abernathy, H. Yang, C. J. Pan, J. . Chyi, P. Bove, H. Lahreche, and J. Thuret, "Si-diffused Ga_N for enhancement-mode Ga_N MOSFET on Si applications," *Journal of Electronic Materials*, vol. 35, no. 4, pp. 685–690, 2006.
- [64] V. Adivarahan, M. Gaevski, A. Koudymov, J. Yang, G. Simin, and M. Khan, "Selectively doped high-power AlGa_N/InGa_N/Ga_N MOS-DHFET," *Electron Device Letters, IEEE*, vol. 28, no. 3, pp. 192–194, 2007.
- [65] Y. Deng, V. Adivarahan, and A. Khan, "Selective doping and optimization of InGa_N channel and InGa_N back-barrier in deep submicron Ga_N heterojunction field effect transistor with a recessed gate," 2007.
- [66] D. Kim and H. K. Baik, "Current conduction mechanism of Si/Ti-based ohmic contacts to n-Ga_N," *Applied Physics Letters*, vol. 77, no. 7, pp. 1011–1013, 2000.
- [67] V. Desmaris, J. . Shiu, C. . Lu, N. Rorsman, H. Zirath, and E. . Chang, "Transmission electron microscopy assessment of the Si enhancement of Ti/Al/Ni/Au Ohmic contacts to undoped AlGa_N/Ga_N heterostructures," *Journal of Applied Physics*, vol. 100, no. 3, 2006.
- [68] F. M. Mohammed, L. Wang, and I. Adesida, "Ultralow resistance Si-containing Ti/Al/Mo/Au Ohmic contacts with large processing window for AlGa_NGa_N heterostructures," *Applied Physics Letters*, vol. 88, no. 21, 2006.
- [69] N. Onojima, N. Hirose, T. Mimura, and T. Matsui, "Effects of Si deposition on AlGa_N barrier surfaces in Ga_N heterostructure field-effect transistors," *Applied Physics Express*, vol. 1, no. 7, pp. 0711011–0711013, 2008.

- [70] N. Onojima, N. Hirose, T. Mimura, and T. Matsui, "Ultrathin AlN/GaN heterostructure field-effect transistors with deposition of Si atoms on AlN barrier surface," *Applied Physics Letters*, vol. 93, no. 22, 2008.
- [71] S. Arulkumaran, T. Egawa, H. Ishikawa, T. Jimbo, and Y. Sano, "Surface passivation effects on AlGaIn/GaN high-electron-mobility transistors with SiO₂, Si₃N₄, and silicon oxynitride," *Applied Physics Letters*, vol. 84, no. 4, pp. 613–615, 2004.
- [72] P. H. Fang, P. Bai, J. H. Kinnier, Z. Huan, and C. C. Schubert, "Temperature dependent formation of microcrystal and amorphous silicon by vacuum evaporation," *Journal of Non-Crystalline Solids*, vol. 59-60, no. PART 2, pp. 819–821, 1983.
- [73] M. Steinert, J. Acker, S. Oswald, and K. Wetzig, "Study on the mechanism of silicon etching in HNO₃-rich HF/HNO₃ mixtures," *Journal of Physical Chemistry C*, vol. 111, no. 5, pp. 2133–2140, 2007.
- [74] H. R. Philipp and E. A. Taft, "An optical characterization of native oxides and thin thermal oxides on silicon," *Journal of Applied Physics*, vol. 53, no. 7, pp. 5224–5229, 1982.
- [75] C. R. Werrett, A. K. Bhattacharya, and D. R. Pyke, "The validity of c1s charge referencing in the xps of oxidised al-si alloys," *Applied Surface Science*, vol. 103, no. 4, pp. 403–407, 1996.
- [76] N. Koshizaki, H. Umehara, and T. Oyama, "XPS characterization and optical properties of Si/SiO₂, Si/Al₂O₃ and Si/MgO co-sputtered films," *Thin Solid Films*, vol. 325, no. 1-2, pp. 130 – 136, 1998.
- [77] M. Matsuoka, S. Isotani, W. Sucasaire, L. Zambom, and K. Ogata, "Chemical bonding and composition of silicon nitride films prepared by inductively coupled plasma chemical vapor deposition," *Surface and Coatings Technology*, vol. 204, no. 18-19, pp. 2923 – 2927, 2010.
- [78] D. Briggs and M. Seah, *Practical Surface Analysis. Volume 1. Auger and X-ray Photoelectron Spectroscopy*. Chichester, UK: John Wiley and Sons, 1990.
- [79] J. Liang, L. Topor, A. Navrotsky, and M. Mitomo, "Silicon nitride: enthalpy of formation of the α - and β -polymorphs and the effect of C and O impurities," *Journal of Materials Research*, vol. 14, no. 5, pp. 1959–1968, 1999.

Appendix A

GaN IC FABRICATION PROCESS FLOW

List of process flow specific abbreviations:

BOE	Buffered Oxide Etchant
DI	Deionized (Water)
HMDS	Hexamethyldisilazane, Adhesion Promoter
ICP	Inductively Coupled Plasma
IPA	Isopropyl Alcohol
RIE	Reactive Ion Etch

A.1 NRC's standard GaN IC Fabrication Process Flow (some information is intentionally omitted)

1. Ohmic Contact Lithography

- Native oxide removal
- Double resist layer coat/bake
- UV exposure: GANFET Ohmic mask
- Develop. Inspect

2. Ohmic Metallization

- Microwave descum
- Pre clean

- E-beam deposition: Ti/Al/Ti/Au ohmic metal stack
- Lift-off
- Rinse: acetone, IPA, DI rinse
- Microwave descum

3. Ohmic Metal Anneal

- Inspect for complete lift-off prior to anneal
- RTA Anneal: 845C, 1 min. in N₂ flow
- Measure: contact resistance and saturation currents

4. Mesa Lithography

- Photoresist coat/bake
- UV exposure: GANFET Mesa mask
- Develop. Inspect

5. Mesa Etch in ICP

- ICP etch
- Resist strip
- Clean: IPA, DI rinse
- Microwave descum/oxide removal
- Characterization: surface profiling with Dektak

6. Gate Metal Optical Lithography

- Native oxide removal
- Double resist layer coat/bake
- UV exposure: GANFET Gate Metal mask
- Develop. Inspect

7. Gate Metallization

- Microwave descum
- Pre clean
- Deposition: Pt based Schottky metal stack
- Lift-off
- Rinse: acetone, IPA, DI rinse
- Microwave descum

8. Gate Metal Anneal

- Inspect for complete lift-off prior to anneal
- RTA Anneal in N₂ flow

9. Silicon Nitride Passivation

- Microwave descum

- Pre-clean
- Deposit SiN layer

10. **Pattern for RIE Etch (VIA1)**

- Resist coat/bake
- UV exposure: GANFET VIA1 mask
- Develop, DI rinse
- Microwave descum. Inspect

11. **Silicon Nitride VIA1 Etch**

- ICP etch
- Resist strip
- Rinse: IPA, DI rinse
- Microwave descum. Inspect

12. **Pattern for 1ME**

- Thermoplastic coat/bake
- Resist coat/bake
- UV exposure: GANFET 1ME Mask
- Resist develop, DI rinse
- Microwave descum
- DUV exposure
- Thermoplastic develop. Inspect

13. **1ME Metallization**

- Microwave descum
- Deposition: Au based 1ME metal stack
- Lift-off
- Rinse: acetone/IPA
- RTA anneal in N2 flow. Inspect

A.2 Batch-1 Fabrication Process Flow

NOTE: ‘*’ denotes a standard NRC’s process step.

1. **a-Si Thin Film Coating**

- Pre clean
- E-beam evaporation: 30 nm of a-Si on designated samples. Use magnets to protect corners for probing access. Include Si test wafer

2. **SiON Thin Film Coating**

- Pre clean

- PECVD: 60 nm of silicon oxinitride (SiON) on designated samples. Include Si test wafer

3. Surface Encapsulation with SiO₂

- PECVD: 100 nm of SiO₂ on thin film coated samples. Include Si test wafer

4. Corner Patterning and Wet Etching for Probing Access

- Clean: Acetone, IPA, DI rinse
- Photoresist coat/bake
- Expose corners manually with microscope light
- Resist develop, DI rinse.
- BOE etch test wafers: determine SiO₂ and SiO₂ + SiON etch rates by observing hydrophobic properties of the Si wafers
- BOE etch: etch SiO₂ and SiO₂ + SiON using measured times. DI rinse
- Resist strip: acetone, IPA, DI rinse

5. Stepwise Heat Treatment Anneals and Hall Measurements

- Initial electrical Hall measurement using access corners
.....
- Clean: acetone, IPA, DI rinse
- RTA Anneal: $T_1 = 400\text{C}$, t_1 min. in N₂ flow
- Electrical Hall measurement using access corners
- Selectively repeat the anneals up to $T_n = 950\text{C}$
.....

6. Wet Etch of SiO₂, a-Si and SiON Films

- Clean: Acetone, IPA, DI rinse
- Use previously determined wet etch duration for SiO₂ and SiON wet etch in BOE. DI rinse. Inspect
- a-Si wet etch in 1:3 HF:HNO₃, 5 sec. DI rinse. Inspect
- Electrical Hall measurement

7. *Ohmic Contact Lithography

8. *Ohmic Metallization

9. *Ohmic Metal Anneal

10. Post Ohmic Anneal Electrical Measurements

- C-TLM ohmic contact and sheet resistance measurements

11. *Mesa Lithography

12. *Mesa Etch in ICP

13. *Gate Metal Optical Lithography

14. *Gate Metallization

15. ***Gate Metal Anneal**
16. **Electrical Measurements, Unpassivated Devices**
 - C-TLM ohmic contact and sheet resistance
 - TLM ohmic contact and sheet resistance
 - HEMT DC and pulsed I-V/ Current collapse
 - HEMT gate Schottky junction I-V
 - HEMT drain-source breakdown voltage
17. ***Silicon Nitride Passivation**
18. ***Pattern for RIE Etch (VIA1)**
19. ***Silicon Nitride VIA1 Etch**
20. **Electrical Measurements, Passivated Devices**
 - C-TLM ohmic contact and sheet resistance
 - TLM ohmic contact and sheet resistance
 - HEMT DC and pulsed I-V/ Current collapse
 - HEMT gate Schottky junction I-V
 - HEMT drain-source breakdown voltage
21. ***Pattern for 1ME**
22. **1ME Metallization**
 - Microwave descum
 - Deposition: Au based 1ME metal stack
 - Lift-off
 - Rinse: acetone/IPA
23. **Post 1ME Electrical Measurements**
 - C-TLM ohmic contact and sheet resistance
 - TLM ohmic contact and sheet resistance

A.3 Batch-2 Fabrication Process Flow

NOTE: ‘*’ denotes a standard NRC’s process step.

1. **a-Si (30 nm) Thin Film Coating**
 - Pre clean
 - E-beam evaporation: 30 nm of a-Si on designated samples. Include Si test wafer
2. **a-Si (5 nm) Thin Film Coating**
 - Pre clean
 - E-beam evaporation: 5 nm of a-Si on designated samples at 1 Å/sec dep. rate. Include Si test wafer

3. Surface Encapsulation with SiO₂

- PECVD: 100 nm of SiO₂ on 30 nm a-Si thin film coated samples. Include Si test wafer

4. Heat Treatment Anneals and Hall Measurements

- RTA Anneal: designated 30 nm a-Si coated samples, 850C, 15 min. in N₂ flow
- RTA Anneal: designated 30 nm a-Si coated samples, 950C, 5 min. in N₂ flow
- RTA Anneal: designated 30 nm a-Si coated samples, 1,000C, 5 min. in N₂ flow

5. Wet Etch of SiO₂ and a-Si Films

- Clean: Acetone, IPA, DI rinse
- BOE etch test wafers: determine SiO₂ etch rates by observing hydrophobic properties of the Si wafers
- Use previously determined wet etch duration for SiO₂ wet etch in BOE. DI rinse. Inspect
- Perform Hall measurement on a new control sample. Add to HF:HNO₃ wet etch batch
- a-Si wet etch in 1:3 HF:HNO₃, 10 sec. DI rinse. Inspect
- Electrical Hall measurement on a new control sample

6. *Ohmic Contact Lithography

7. *Ohmic Metallization

8. Stepwise Ohmic Metal Anneal

- Inspect for complete lift-off prior to anneal
.....
- RTA Anneal: $T_1 = 750\text{C}$, 1 min. in N₂ flow
- Measure: c-TLM contact and sheet resistance, saturation currents over 2 μm c-TLM gap
- Repeat the anneals for increasing temperatures until the best ohmic contacts are achieved
.....

9. SIMS and XPS Measurements

- Perform SIMS profiling on dedicated samples
- Perform XPS chemical composition measurement on dedicated samples

A.4 Batch-3 Fabrication Process Flow

NOTE: “*” denotes a standard NRC’s process step.

1. a-Si (30 nm) Thin Film Coating

- Pre clean
- E-beam evaporation: 30 nm of a-Si on designated samples. Include Si test wafer

2. a-Si (5 nm) Thin Film Coating

- Pre clean
- E-beam evaporation: 5 nm of a-Si on designated samples at 1 Å/sec dep. rate. Include Si test wafer

3. Surface Encapsulation with SiO₂

- PECVD: 100 nm of SiO₂ on 30 nm a-Si thin film coated samples. Include Si test wafer

4. Heat Treatment Anneals and Hall Measurements

- RTA Anneal: 30 nm a-Si coated samples, 950C, 5 min. in N₂ flow

5. Wet Etch of SiO₂ and a-Si Films

- Clean: Acetone, IPA, DI rinse
- BOE etch test wafers: determine SiO₂ etch rates by observing hydrophobic properties of the Si wafers
- Use previously determined wet etch duration for SiO₂ wet etch in BOE. DI rinse. Inspect
- a-Si wet etch in 1:3 HF:HNO₃ 10 sec. DI rinse. Inspect

6. XPS Measurements

- Perform XPS chemical composition measurement on dedicated samples

7. Surface Contact Angle Measurement

- Measure contact angle
- Clean: Acetone, IPA, DI Rinse. Measure contact angle
- Clean: Acetone, scrub with IPA, DI rinse. Measure contact angle
- HMDS prime. Measure contact angle
- Photoresist coat/bake. Develop. DI rinse. Measure contact angle
- Microwave descum. Measure contact angle
- Standard ohmic met. dep. pre clean. DI rinse. Measure contact angle
- DI rinse. Measure contact angle
- Store samples for ~24 hours. Measure contact angle

8. XPS Measurements

- Perform XPS chemical composition measurement on dedicated samples

9. *Ohmic Contact Lithography

10. *Ohmic Metallization

11. Stepwise Ohmic Metal Anneal

- Inspect for complete lift-off prior to anneal
.....
- RTA Anneal: $T_1 = 825\text{C}$, 1 min. in N₂ flow
- Measure: c-TLM contact and sheet resistances, saturation current over 2 μm c-TLM gap

- Repeat the anneals for increasing temperatures until best ohmic contacts are achieved
.....

12. ***Mesa Lithography**

13. ***Mesa Etch in ICP**

14. ***Gate Metal Optical Lithography**

15. ***Gate Metallization**

16. ***Gate Metal Anneal**

17. **Electrical Measurements, Unpassivated Devices**

- TLM ohmic contact and sheet resistance
- HEMT DC and pulsed I-V / Current collapse
- HEMT transconductance and threshold voltage

18. ***Silicon Nitride Passivation**

19. ***Pattern for RIE Etch (VIA1)**

20. ***Silicon Nitride VIA1 Etch**

21. ***Pattern for 1ME**

22. ***1ME Metallization**

23. **Electrical Measurements, Passivated Devices**

- C-TLM/TLM ohmic contact and sheet resistance, current density over 2 μm c-TLM gap
- HEMT DC and pulsed I-V / Current collapse
- HEMT transconductance and threshold voltage
- HEMT gate Schottky junction I-V
- HEMT gate leakage versus drain voltage for standard and field plate assisted devices