Planar and non-gold metal stacks processes and conduction mechanisms for AlGa_N/Ga_N high-electron-mobility transistors on silicon

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2017

Li, Y. (2017). Planar and non-gold metal stacks processes and conduction mechanisms for AlGa_N/Ga_N high-electron-mobility transistors on silicon. Doctoral thesis, Nanyang Technological University, Singapore.

<http://hdl.handle.net/10356/72250>

<https://doi.org/10.32657/10356/72250>

**NANYANG
TECHNOLOGICAL
UNIVERSITY**

**PLANAR AND NON-GOLD METAL STACKS
PROCESSES AND CONDUCTION MECHANISMS
FOR ALGAN/GAN HIGH-ELECTRON-MOBILITY
TRANSISTORS ON SILICON**

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2016

**Planar and Non-gold Metal Stacks Processes and
Conduction Mechanisms for AlGaN/GaN High-
Electron-Mobility Transistors on Silicon**

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School of Electrical & Electronic Engineering

A thesis submitted to the Nanyang Technological University
in partial fulfillment of the requirements for the degree of

Doctor of Philosophy

2016

Statement of Originality

I hereby certify that the work embodied in this thesis is the result of original research and has not been submitted for a higher degree to any other university or institution.

Date

Li Yang

Acknowledgements

When I look back to recollect the past years I have spent in the Nanyang Technological University, I realize that it is of my great honor to join the GaN HEMT group and work with so many outstanding researchers closely. This part of experience has become an unforgettable memory of my life. Before presenting the work of this thesis, I want to acknowledge my appreciations to all individuals who have helped me during my Ph.D. study.

First and foremost, I would like to express my deepest gratitude to my supervisor Associate Professor Ng Geok Ing for his continuous encouragements and support to my study. I have benefited a lot from his knowledge and experience in the areas of III-Nitrides. His foresight always guided me in the right direction. His rigorous attitude towards research works impressed me very deeply. Undoubtedly, what I have learnt from Prof Ng will become the precious treasures of my life and will benefit my future work.

Next, I would like to present my genuine thankfulness to my co-supervisor Dr. Subramaniam Arulkumaran. Dr. Arul is so knowledgeable and could offer me many detailed instructions for my research work. In addition, he is always patient in guiding me with a lot of face-to-face discussions. The experience with Dr. Arul will certainly be great helpful to my future career.

I would like to express my sincere appreciation to Associate Professor Wang Hong, Ang Kian Siong, Dr. Liu Zhi Hong, and Dr. Liu Chong Yang for their help in my work. Prof Wang has given me insightful ideas in the early

years of my PhD study. Kian Siong has offered me his assistance in device fabrication. Zhi Hong and Chong Yang have given me inspirations as well as valuable hand-on trainings, by which my research work was facilitated significantly.

My special thanks to Professor Zhang Dao Hua and Dr. Xu Zheng Ji. Prof Zhang allowed me to use the sputter system in his group and assigned Dr. Xu to train me on the system operation.

In addition, I would like to thank all my team members. They are Dr. M. J. Anand, Dr. Ye Gang, R. Hofstetter, Wei Chuan, Ze Cen, K. Ranjan, Matthew, and Abhinay. I am also grateful to the MTDC team members Manoj, Siu Chuen, Louis, Darren, and Vicknesh for their help in this work.

I will not forget the contributions of the technical staff from the Characterization lab, the Silicon Technologies Center of Excellence (Si-COE), and the Nanyang Nanofabrication Center (N2FC). Many thanks to Fauzi, Seet, Kowk Fai, Shamsul, Gang Yih, Mak, Xiao Hong, Ling Ling, Irene, Tina and Tong Sun for their support.

Also, I want to present my gratitude to my friends and colleagues from the School of Electrical and Electronic Engineering: Dr. Wang Xing Hui, Dr. Cheng Yuan Bin, Dr. Li Cheng Ming, Dr. Sun Lei Meng, Hu Xiao Nan, Dr. Yu Xue Chao, Li Xiang, Dr. Qiao Zhong Liang, Dr. L. Ravikiran, Dr. Meng Bo, Bao Shu Yu, and Yan Zhi Yu, for their support in past years.

Finally and most importantly, I want to dedicate this work to my mother. As the only son of my parents, I am supposed to stay with them and take care of

them in their elder ages. It is not an easy decision for my mother to support my study in another country, but she has never in one moment hesitated in helping to realize my dream to complete my Ph.D. I am so lucky to be her son.

Abstract

AlGaIn/GaN high-electron-mobility transistors (HEMTs) have demonstrated great potentials in high voltage and high frequency applications due to the advantage in their intrinsic material properties such as wide band gap, high electron saturation velocity, high sheet carrier density, etc. Conventionally, AlGaIn/GaN HEMTs are realized in III-V fabrication line using the processes such as mesa isolation, gold-based ohmic contacts and gate electrodes etc. that are not used in Si CMOS line

With the advances of epitaxial growth technology, the crack-free AlGaIn/GaN heterostructure can be realized in large diameter (200 mm) Si substrate with low wafer bowing values, which provides the possibility of fabricating devices on larger size and lower cost substrates in contrast to silicon carbide substrates. In recent years, CMOS-compatible processes have received considerable attentions. This approach allows high-volume low-cost fabrication of AlGaIn/GaN HEMTs by utilizing the existing Si foundries. Also, the CMOS-compatible processes offer the opportunity of on-chip integration of GaN devices with Si integrated circuits (ICs).

In this thesis, the focus is to develop the key process modules which will be CMOS compatible for AlGaIn/GaN HEMTs. The three process modules include planar isolation by ion-implantation, non-gold ohmic contact, and non-gold Schottky gate. In addition, the conduction mechanisms and its physical origin are also systematically analyzed to understand the impact of developed

processes to the characteristics of AlGaIn/GaN HEMTs on Si substrate.

The planar isolation is achieved in AlGaIn/GaN HEMTs using multi-energy ultra-heavy $^{131}\text{Xe}^+$ ion-implantation. About an order of magnitude lower buffer leakage current, and an order of magnitude higher buffer sheet resistance and ON/OFF current ratio are observed in the implant-isolated HEMTs when compared to the conventional mesa-isolated HEMTs. The Arrhenius plot reveals an activation energy (0.513 eV) that belongs to the lattice damage caused by heavy ions, which is the origin of the high resistivity in the implanted region. The implant-isolated AlGaIn/GaN HEMTs exhibit comparable DC output and transfer characteristics to the mesa-isolated HEMTs. Due to the elimination of leakage paths at the mesa side wall, the buffer breakdown voltage is almost unchanged with and without Si_3N_4 passivation. The OFF-state breakdown voltage is also improved and the gate leakage current is reduced.

The non-gold ohmic contacts are realized by Ta/Si/Ti/Al/Ni/Ta metal stacks with a low contact resistance ($\sim 0.22 \text{ } \Omega\cdot\text{mm}$) and a smooth surface morphology. The low contact resistance is attributed to the formation of low work function Titanium Silicides (TiSi_x) at the metal-semiconductor (M-S) interface. As the AlGaIn/GaN epi-layers are protected by the bottom-Ta refractory layer, the metal stacks do not react with (Al)GaN. Hence, the conduction mechanism of AlGaIn/GaN HEMTs with our non-gold metal stacks is confirmed to be thermionic emission (TE) through a low energy barrier ($\sim 0.113 \text{ eV}$) at the M-S interface, which is different from the thermionic-field emission (TFE) observed in conventional gold-based ohmic contacts.

The non-gold Schottky gate is realized by sputtered TiN with an improved Schottky barrier height (SBH ~ 1.1 eV) as compared to other reported TiN-based gates. The improvement of SBH may be due to the incorporation of oxygen, the surface crystal orientation, or the relatively high N₂ ratio in the total gas flow. The study on gate leakage current reveals that the conduction mechanism is dominated by Poole-Frenkel emission at low reverse bias (-3.2 V $< V_R < 0$) through an interface state of 0.53 eV, and by trap-assisted tunneling at high reverse bias (-20 V $< V_R < -3.2$ V) through another interface state of ~ 0.115 eV. The AlGaIn/GaN HEMTs with sputtered TiN gate demonstrate comparable DC output and transfer characteristics and dynamic/static ON-resistance ratio, improved OFF-state breakdown voltage, and reduced gate leakage current as compare to the device with conventional Ni/Au gate.

In summary, three important and challenging process modules in developing CMOS-compatible processes for the fabrication of AlGaIn/GaN HEMTs on Silicon substrate have been systematically studied and reported in this thesis. The multi-energy ultra-heavy ¹³¹Xe⁺ implantation, ohmic contacts with non-gold metal stacks (Ta/Si/Ti/Al/Ni/Ta), and non-gold Schottky gate (sputtered TiN) are demonstrated with improved device performance. The conduction mechanisms of both ohmic and Schottky contacts are also systematically studied. The findings in this thesis provide valuable solutions for the development of CMOS-compatible processes, which can be potentially used for the high-volume and low-cost fabrication of AlGaIn/GaN HEMTs in Si foundries for commercial applications.

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Chapter 1 Introduction

1.1 Overview of AlGaN/GaN high-electron-mobility transistors and power switching applications

AlGaN/GaN high-electron-mobility transistors (HEMTs) have exhibited excellent capabilities in high power [1], high frequency [2][3], high voltage [4][5] and high temperature operations [6] due to their excellent electronic properties of GaN. Table 1-1 gives a detailed comparison of GaN with Si, GaAs, and 4H-SiC in material properties. Owing to the large band gap and the high breakdown electric field, GaN is suitable for the fabrication of high voltage devices, such as Schottky barrier diodes (SBDs) [7] and HEMTs. Furthermore, the superior electron mobility and saturated drift electron velocity in AlGaN/GaN heterostructure enable the HEMTs to operate at very high frequencies. The large values of Johnson's figure of merit (FOM) and Baliga's FOM suggest the attractive capabilities of AlGaN/GaN HEMTs in applications which require high output power with high frequency [8] and low-loss at low frequency [9]. Figure 1-1 gives an overview of the switching capacity as a function of operation frequency. AlGaN/GaN HEMTs are capable to cover a wide range of frequencies which are currently dominated by other technologies such as Si gate turn-off thyristor (GTO), insulated gate bipolar transistor (IGBT), metal-oxide-semiconductor field effect transistor (MOSFET), SiC MOSFET, etc.

Table 1-1 Material properties of GaN as compared to other semiconductor materials of interest

<i>Material</i>	<i>Si</i>	<i>GaAs</i>	<i>4H-SiC</i>	<i>GaN</i>
E_g (eV)	1.12	1.42	3.26	3.42
ϵ_s	11.8	12.9	10.0	9.0
μ_n ($cm^2/V \cdot s$)	1350	8500	700	900-2000
E_c ($10^6 V/cm$)	0.3	0.4	3	3.3
$v_s(v_p)(10^7 cm/s)$	1.0 (1.0)	1.3 (2.1)	2.0 (2.0)	1.5 (2.1)
$\kappa(W/cm \cdot K)$	1.5	0.54	4.5	1.3-2.0
Johnson's FOM	1	7.1	180	760
Baliga's FOM	1	15.6	130	650

E_g - band gap; ϵ_s - dielectric constant; μ_n - electron mobility;

E_c - breakdown electric field; v_s (v_p) - peak drift velocity of electron (hole);

κ - thermal conductivity;

Figure 1-1 The switching capacity as function of the operation frequency for different device categories

In addition to switching applications, the demand for GaN power devices is also increasing rapidly. Figure 1-2 shows the wide band gap (WBG) power device market from 2014 to 2020 according to the report by Yole Développement (website: www.yole.fr). With respect to the compound annual growth rate (CAGR) of 22% for SiC power device, the market of GaN power device is expanding with a more aggressive CAGR of 95%. In 2020, the GaN power device market is estimated to be \$303 M in a \$12.7 B overall device market.

Figure 1-2 Wide band gap (WBG) power device market from 2014 to 2020

The power switching AlGaN/GaN HEMTs with state-of-the-art on-resistance (R_{on}) and OFF-state breakdown voltage (BV_{OFF}) values are summarized in Figure 1-3. The data points in Figure 1-3 are chronologically listed in Table 1-2. Researchers continuously strived for better device performance with higher BV_{OFF} and lower R_{on} in the past decades. To date, the

largest BV_{OFF} of 10400 V (with $R_{on} \sim 186 \text{ m}\Omega\cdot\text{cm}^2$) was reported by Y. Uemoto, et al. through the integration of poly-AlN passivation, deep via-holes through sapphire, and field-plate processes [10]. By reducing the source-drain space to $2.2 \text{ }\mu\text{m}$, an extremely small R_{on} of $0.09 \text{ m}\Omega\cdot\text{cm}^2$ (with $BV_{OFF} \sim 35 \text{ V}$) has been realized by M. Inada, et al [11].

Figure 1-3 State-of-the-art power switching AlGaN/GaN HEMTs from year 2000 to 2016, with the performance limit of GaN, 4H-SiC, GaAs and Si

Table 1-2 Reported OFF-state breakdown voltage (BV_{OFF}), on-resistance (R_{on}), and power figure of merit (Power-FOM) values of state-of-the-art power switching AlGaN/GaN HEMTs from year 2000 to 2016

Points	Year	Substrate	BV_{OFF} (V)	R_{on} ($m\Omega \cdot cm^2$)	Power-FOM ($V^2/\Omega \cdot cm^2$)	Ref
1	2000	sapphire	570	1.5	2.1×10^8	[12]
2	2001	SiC	1050	3.4	3.2×10^8	[13]
3	2003	SiC	600	3.3	1.1×10^8	[14]
4	2004	sapphire	900	2.8	2.9×10^8	[15]
5	2005	insulating	600	0.6	6×10^8	[16]
6	2006	sapphire	1600	3.4	7.5×10^8	[17]
7	2006	SiC	1900	2.2	1.6×10^9	[18]
8	2007	SiC	480	2.1	1.1×10^8	[19]
9	2007	Si	650	3.1	1.4×10^8	[20]
10	2007	sapphire	8300	186	3.7×10^8	[21]
11	2009	Si	1000	5	2×10^8	[22]
12	2009	Si	1300	6.2	2.7×10^8	[23]
13	2009	Si	830	6.2	1.1×10^8	[24]
14	2009	sapphire	10400	186	5.8×10^8	[10]
15	2010	Si	580	1.25	2.7×10^8	[25]
16	2010	Si	643	4.3	9.6×10^7	[26]
17	2010	Si	305	0.45	2.1×10^8	[27]
18	2010	Si or glass	1500	5.3	4.2×10^8	[28]
19	2010	SiC	942	0.39	2.3×10^9	[29]
20	2010	SiC	700	0.68	7.2×10^8	[30]
21	2011	Si	650	2.0	2.1×10^8	[31]
22	2012	Si	1800	1.9	1.7×10^9	[32]
23	2012	Si	1400	9.6	2.6×10^8	[33]
24	2012	Si	1590	1.86	1.4×10^9	[34]
25	2013	Si	1100	2.3	5.3×10^8	[35]
26	2013	SOI	1471	3.92	5.5×10^8	[36]
27	2013	Si	640	0.92	4.5×10^8	[37]
28	2013	Si	632	1.3	3.1×10^8	[38]
29	2014	Si	1900	1.6	2.3×10^9	[5]
30	2014	bulk-GaN	1500	2.2	1.0×10^9	[39]
31	2015	Si	3000	10.1	8.9×10^8	[40]
32	2016	Si	2400	6.5	8.9×10^8	[41]

1.2 GaN-on-Si

Owing to the unavailability of large size bulk GaN substrate as well as the high cost and the difficulties of synthesis, AlGaN/GaN heterostructures are typically grown on non-native substrates such as sapphire (Al_2O_3), SiC and Si (111). The properties of commonly used substrates are listed in Table 1- 3.

Table 1-3 Properties of commonly used substrates for GaN growth

<i>Material</i>	<i>Al_2O_3 (100)</i>	<i>6H-SiC</i>	<i>GaN (0001)</i>	<i>Si (111)</i>
a (Å)	4.758	3.081	3.189	3.846
a mismatch	-33%	3.5%	0	-17%
α (10^{-6}K^{-1})	7.3	4.5	5.6	2.6
α mismatch	-23%	24%	0	116%
κ (W/cm·K)	0.5	4.5	1.3	1.5
Max. D (mm)	200	150	50	300
Cost	Medium	Very high	Extremely high	Low

a - Lattice constant; α - Thermal expansion coefficient;

κ - Thermal conductivity; D - substrate diameter.

Attracted by the availability of large-diameter Si substrates with low cost and the maturity of Si CMOS technology, the growth of AlGaN/GaN heterostructure on Si has received increasing interests over the past years. Many difficulties that are associated with the mismatch of the lattice constant and the thermal expansion coefficient between GaN and Si have been resolved. For example, high dislocation density and wafer bowing can be solved using transition layers (TLs) to alleviate the stress in between Si substrate and

epitaxial GaN. By inserting TLs such as graded $\text{Al}_x\text{Ga}_{1-x}\text{N}$ buffer layer [42], GaN/AlN superlattice structure [43], single $\text{Al}_x\text{Ga}_{1-x}\text{N}$ with low Al mole fraction [44], AlN [45] or periodic SiN [46], etc., high quality GaN epitaxial layers on Si has been realized for the fabrication of AlGaN/GaN HEMTs.

Impressive achievements have been reported for the high performance AlGaN/GaN HEMTs on Si in the past 15 years. In 2012, M. Sun, et al. demonstrated the AlGaN/GaN HEMT device on Si with BV_{OFF} of 1800 V, R_{on} of $1.9 \text{ m}\Omega\cdot\text{cm}^2$ and power-FOM of $1.7\times 10^9 \text{ V}^2/\Omega\cdot\text{cm}^2$ using N^+ implantation [32]. Lately, N. Herbecq, et al. further pushed BV_{OFF} to 3000 and 2400 V but with increased R_{on} of 10.1 and $6.5 \text{ m}\Omega\cdot\text{cm}^2$, respectively [40][41]. In high frequency operation, the maximum oscillation frequency (f_{max}) of 206 GHz [47] and the cut-off frequency (f_{T}) of 153 GHz [48] have been reported for the AlGaN/GaN HEMTs on Si. Although the improvement on output power density was hindered by poor thermal conductivity of Si substrate, some good power results, e.g. 2.5 W/mm at 40 GHz [49], 5.1 W/mm at 18 GHz [50], 8.1 W/mm at 8 GHz [51] and 12 W/mm at 2.14 GHz [52], have been demonstrated for AlGaN/GaN HEMTs on Si.

1.3 CMOS-compatible process for AlGaN/GaN HEMT fabrication

Traditionally, most of the AlGaN/GaN HEMTs on Si with state-of-the-art performances were fabricated in conventional III-V processing line. These

processes are very different from those used in standard Si fabrication line. For example, non-planar inter-device isolation is achieved by mesa recess using plasma etching [48][51], and both the Schottky gate [32][40][41][48][51][52] and the ohmic contacts [32][40][41][47][48][51][52] are formed by gold-based metal scheme. Both Mesa isolation and gold based contacts are not allowed in Si CMOS fab as the mesa etching will yield a non-planar wafer surface and the usage of gold will inevitably contaminate the Si CMOS processing line. In recent years, the development of CMOS-compatible processes received a lot of attentions mainly due to the following factors:

- (1) The high performance GaN-based electronics with low cost are demanded in the expanding commercial market. The epitaxial growth of high quality AlGaIn/GaN HEMT structure on large-diameter (e.g. 200 mm [35][53]) Si substrates offers the possibility of fabricating AlGaIn/GaN HEMTs with high throughput in existing Si foundries, which could not only save in the heavy capital investment but also reduce the production cost. Hence, the development of CMOS-compatible processes for AlGaIn/GaN HEMTs fabrication on 200 mm Si CMOS foundries has great potential from the economic point of view.
- (2) In addition, the Si substrate in the GaN-on-Si structure provides the opportunity of on-chip integration of GaN and the well-matured Si electronics [54][55]. The integration of AlGaIn/GaN HEMTs with Si transistors could provide a new platform for the co-existence of analog and mixed-signal on the same chip. To achieve this goal, the CMOS-compatible

processes for the fabrication of AlGaIn/GaN HEMTs is also needed.

Some achievements have been demonstrated on the development of CMOS-compatible processes. For example, the IMEC group, Belgium has reported a fully CMOS-compatible process for the fabrication of AlGaIn/GaN SBDs [7], HEMTs [56] and Metal-insulator-semiconductor HEMTs (MIS-HEMTs) [56][57]. The ohmic contacts were Ti/Al-based stacks with various cap layers (e.g. W [57] and Ti/TiN [7][56]). The gate electrode employed was W [57] or TiN [7][56] as the high work function material. The conventional metal lift-off process has been replaced by metal etch back process using plasma dry etching and the inter-device isolation was achieved by N⁺ implantation for a planar surface morphology [7][56][57]. The MIT group has reported AlGaIn/GaN HEMTs fabricated by Au-free technology [58][59]. Au-free metal stacks were implemented for both ohmic contacts and gates (Ohmic: Ti/Al/Ni/Pt [58] and Ti/Al/W with AlGaIn barrier recess [59]; Gate: Ni/Al [58] and Ti/Al/W [59]). Other research groups have also reported CMOS-compatible process modules, such as non-gold ohmic contact [60][61], non-gold Schottky gate [62], and isolation by ion-implantation [63][64].

1.4 Motivation and Objective

Although significant advances have been reported on the development of CMOS-compatible processes for AlGaIn/GaN HEMTs fabrication, more works can be done to further improve the performances. For planar isolation, the

reported implantation using N^+ [65], P^+/He^+ [66], and O^+ [67], did not exhibit better isolation effect than conventional mesa isolation. The sheet resistance of the implanted region was observed to decrease at high temperature annealing. Hence, the planar isolation by alternative heavier ions with improved thermal stability needs to be explored.

For the non-gold ohmic contact, the reported contact resistance (R_c) values are still relatively high compared to gold-based ohmic contacts (e.g. $R_c \sim 0.49 \Omega \cdot \text{mm}$ [59], $\sim 0.65 \Omega \cdot \text{mm}$ [57], compared to $R_c < 0.2 \Omega \cdot \text{mm}$ [68][69] for gold-based ohmic contacts). Hence, the further optimization of non-gold ohmic contacts with lower R_c is needed. Similarly, most of the reported non-gold Schottky gate did not exhibit larger Schottky barrier height (SBH or $q\Phi_B$) than conventional Ni/Au gate. Studies for the improvement of $q\Phi_B$ are necessary. Finally and more significantly, the investigations on the CMOS-compatible processes mainly focused on the performance of device. The physical mechanisms, such as the conduction mechanism of non-gold ohmic contact and non-gold Schottky gate, have not been reported thus far.

This thesis focuses on three key process modules which are most challenging in the development of CMOS-compatible processes. These are planar isolation by ion implantation, non-gold ohmic contact, and non-gold Schottky gate. Moreover, detailed studies are carried out to investigate their physical transport mechanisms in order to provide a better understanding of the impact of these developed CMOS-compatible processes on the device performance.

In chapter 2, the basic material properties of III-nitride and $\text{Al}_x\text{Ga}_{1-x}\text{N}$ and operation principles of AlGaN/GaN HEMTs will be briefly introduced. Then, the conventional processing flow for AlGaN/GaN HEMT in III-V fab will be explained and compared with the Si CMOS fabrication technologies. Three major challenges, including planar device isolation, non-gold ohmic contact, and non-gold Schottky gate, will be proposed and discussed.

In chapter 3, the mesa isolation as well as other reported isolation technologies will be reviewed. Next, a comparative study of the isolation effect between multi-energy $^{131}\text{Xe}^+$ implantation and conventional mesa etching will be presented. This is then followed by an analysis on the origin of activation energy. Finally, the characteristics of HEMTs isolated by implantation are demonstrated and compared with the devices isolated by mesa etching.

In chapter 4, the development history of ohmic contacts, which includes conventional gold-based and CMOS-compatible non-gold contacts, will be reviewed initially. Next, the Transmission Line Model (TLM) measurement will be introduced. The non-gold Ta/Si/Ti/Al/Ni/Ta ohmic contact used in this study will be presented in the subsequent sections. The discussion consists of the characterization of ohmic contacts, the analysis on reduced R_c , and the study on conduction mechanism.

In chapter 5, the conventional Ni/Au bi-layer gate and other reported non-gold gates on AlGaN/GaN HEMT will be reviewed. The investigation of sputtering TiN Schottky gate will focus on the characterization of SBH by various methods, the analysis on improved SBH, and the demonstration of

characteristics of AlGaN/GaN HEMT with sputtered TiN gate. Finally, the gate leakage current mechanism will be systematically studied.

In chapter 6, the conclusions and the key contributions of the thesis will be summarized. Recommendations for the future research scope will also be presented.

Chapter 2 AlGa_N/Ga_N HEMTs: Operating Principles and Fabrication Process

Due to the advantages in the intrinsic materials properties of GaN, AlGa_N/Ga_N HEMTs exhibit great potentials in high voltage and high frequency applications. In this chapter, a brief overview of the material properties of III-nitride and the operating principles of AlGa_N/Ga_N HEMTs will be given. Next, basic device characteristics such as DC output, DC transfer and pulsed I-V characteristics which are used in this work to evaluate the fabricated devices will be discussed. In section 2.4, conventional AlGa_N/Ga_N HEMTs fabrication processes, which are adopted in this thesis work for the fabrication of reference HEMTs for the comparison of device performance, will be presented. Finally, the incompatibility of conventional III-V processes with Si fabrication line will be discussed. The alternative approaches for the three key process modules in AlGa_N/Ga_N HEMTs namely device isolation, ohmic contact and Schottky gate to enable a Si CMOS-compatible process will also be highlighted.

2.1 Material properties of III-nitrides and Al_xGa_{1-x}N

Figure 2-1 shows the band gap energy versus lattice constant of various kinds of semiconductors [70]. Compared to other semiconductor materials such as InP, GaAs and Si, III-nitrides exhibit apparently smaller lattice constant in which stronger bonding energy between constituent atoms exists thus resulting

in an extremely high chemical stability. Moreover, this stronger bonding energy and the small atomic mass significantly reduce the lattice scattering, and hence bring high thermal conductivity and high electron saturation velocity. Owing to the wide band gap, III-nitride compounds are able to withstand higher electric field and generate fewer intrinsic carriers even at high temperature. Based on the above stated materials properties, III-Nitride compound semiconductors show great potential in high power [1], high frequency [2], high voltage [4] and high temperature [6] applications.

Figure 2-1 Relation between bond length and band gap energy for various kinds of semiconductors [70].

The basic material properties of GaN along with other semiconductors, such as Si, GaAs, and 4H-SiC etc., are listed in Table 1-1 of Chapter 1. When Al atoms are intentionally incorporated into GaN during epitaxial growth, which partially substitute the Ga atoms, the ternary compound $\text{Al}_x\text{Ga}_{1-x}\text{N}$ is

obtained, where x denotes the mole fraction of Al. The basic material properties of $\text{Al}_x\text{Ga}_{1-x}\text{N}$ as a function of x are given in Table 2-1. With the variation of x from 0 to 1, the change of lattice constant (a_0) is much slower (3.189 nm for α -GaN to 3.112 nm for α -AlN) than the change of the band gap (3.4 eV for GaN to 6.2 eV for AlN) (See Figure 2-1) [71]. From Table 2-1, the incorporation of Al atoms into GaN crystal results in a larger band gap, a higher breakdown field and a larger low-field electron mobility value, as compare to GaN. On the contrary, the lattice constant and the effective electron mass reduce accordingly.

Table 2- 1 Material properties of $\text{Al}_x\text{Ga}_{1-x}\text{N}$ as a function of Al mole fraction x

Properties	Equations	Units	Ref
$E_g, \text{Al}_x\text{Ga}_{1-x}\text{N}$	$3.43 + 1.44x + 1.33x^2$	eV	[72]
$m_n, \text{Al}_x\text{Ga}_{1-x}\text{N}/m_0$	$0.35 - 0.13x$	(n.a.)	[73][74]
$a_0, \text{Al}_x\text{Ga}_{1-x}\text{N}$	$3.189 - 0.077x$	$\text{\AA}$	[71][75]
$\kappa_{\text{Al}_x\text{Ga}_{1-x}\text{N}}$	$[\frac{1-x}{1.3} + \frac{x}{2.5} + \frac{x(1-x)}{6.95}]^{-1}$	$\text{W/m}\cdot\text{K}$	[73][76]
$\mu_{\text{Al}_x\text{Ga}_{1-x}\text{N}}$	$[\frac{x}{1100} + \frac{1-x}{900} + \frac{x(1-x)}{40}]^{-1}$	$\text{cm}^2/\text{V}\cdot\text{s}$	[73][76]
$v_s, \text{Al}_x\text{Ga}_{1-x}\text{N}$	$2.5 - 4.55x + 3.85x^2$	10^7cm/s	[73][76]
$E_c, \text{Al}_x\text{Ga}_{1-x}\text{N}$	$3.3 + 8.45x - 0.05x^2$	10^6V/cm	[73][76]

m_n - effective electron mass; a_0 - lattice constant; μ - low field electron mobility;

2.2 Operating Principles of AlGaIn/GaN HEMTs

The high-electron-mobility transistor (HEMT), which is also named as

heterostructure field-effect transistor (HFET), modulation doped field-effect transistor (MODFET), two-dimensional electron gas field-effect transistor (TEGFET), or selectively doped heterostructure transistor (SDHT), was first invented in 1980 using GaAs/n-AlGaAs heterostructure [77]. The HEMT heterostructure is obtained by epitaxial growth of a larger band gap semiconductor material on top of another smaller bandgap semiconductor material. When the two materials have different band gaps, the 2-dimensional-electron-gas (2DEG) is formed in a triangular quantum well at the interface due to the conduction band offset. For conventional III-V semiconductors such as AlGaAs/InGaAs etc., the materials with larger band gap (typically with doping) is the source of 2DEG. Due to the potential barrier at the hetero-interface, the electrons are isolated from the ionized donors and totally confined in the two dimensional channel. Hence, a high mobility value is obtained because the Coulomb scattering effect between the electron and the ionized donors is significantly reduced, resulting in higher channel current with fast switching speed [77].

The origin of 2DEG in AlGaN/GaN HEMT structure is quite different from that in AlGaAs/GaAs HEMT structure [78]. Owing to the ~ 5 times stronger spontaneous and piezoelectric polarizations effect than GaAs [79], the conductive channel with a high-density of 2DEG can be achieved even without intentional doping in the wide band gap AlGaN for GaN-based heterostructure [80]. The formation of 2DEG in the AlGaN/GaN heterostructure is illustrated in

Figure 2-2. Due to the difference of piezoelectric and spontaneous polarization between the AlGaN barrier and GaN channel, positive charges will be generated at the interface. Consequently, the positive charges attract electrons hence lead to the accumulation of electrons at the interface form a 2DEG channel. Even though the AlGaN barrier is undoped, 2DEG density as high as $\sim 10^{13} \text{ cm}^{-2}$ can be achieved in AlGaN/GaN heterostructure due to the spontaneous and piezoelectric polarization field effects [81].

Figure 2-2 Energy band diagram of two materials with different band-gaps

before and after Fermi level alignment

The principle of gate control in AlGaN/GaN HEMT device is demonstrated in Figure 2-3. Due to the spontaneous and piezoelectric polarization, 2DEG at the triangular quantum well is present even at zero gate bias (see Figure 2-3 (a)). When a positive voltage is applied to the drain, electron flow in the channel will occur due to the potential difference of source

and drain. If the applied gate bias is positive, as shown in Figure 2-3 (b), more electrons are accumulated in the channel and the 2DEG density is enhanced. Consequently a larger drain current will be observed. When the applied gate bias turns negative, the channel region under the gate gets depleted which can reduce the flow of electrons in the channel. With more negative gate bias, the 2DEG beneath the gate will be completely depleted, as shown in Figure 2-3 (c). Except for a very small leakage current (typically 5 - 6 orders of magnitude smaller), the drain current is not observable when the conductive channel is pinched-off.

The AlGaIn/GaN HEMT described in Figure 2-3 is a normally-on (also named as Depletion-mode or D-mode) device. On the contrary, the normally-off (also named as Enhancement-mode or E-mode) HEMT represents the device with a pinched-off channel even no gate bias is applied. Compare to the D-mode device, the E-mode device does not need negative voltage supply. Hence, the circuit designs can be greatly simplified with improved system reliability [82]. In addition, the digital IC application requires the integration of both E-mode and D-mode devices [83]. The reported technologies to realize the E-mode HEMT include fluorine-based plasma treatment [84], recess of AlGaIn barrier [85] (with gate dielectric [82]), p-type injection gate [86], p-n junction gate [83], etc. In this work, only D-mode AlGaIn/GaN HEMT will be discussed.

Figure 2-3 Band diagram of AlGaN/GaN HEMT device at (a) zero gate bias;
(b) positive gate bias; and (c) negative gate bias

2.3 Basic Device Characteristics of AlGaN/GaN HEMTs

Figure 2-4 gives a schematic drawing of a D-mode HEMT device. The geometrical parameters of a HEMT includes: gate-length (L_g), source-gate space (L_{sg}), gate-drain space (L_{gd}), and gate width (W_g). The source-drain current is directly proportional to W_g .

Figure 2-4 Schematic drawing of AlGaN/GaN HEMT device

The ideal DC output and transfer characteristics of AlGaN/GaN HEMT device are shown in Figure 2-5. The dotted line separates the output curves into the linear operation region and the saturation operation region. For high power operation of devices, the large value of maximum source-drain current ($I_{ds, \max}$), which is typically limited by the 2DEG density and the carrier mobility, is

Figure 2-5 Typical DC (a) output and (b) transfer characteristics of AlGaN/GaN HEMTs

required. V. Kumar, et al. reported an $I_{DS, \max}$ of 1.23 A/mm by inserting a Si-doped $\text{Al}_{0.25}\text{Ga}_{0.75}\text{N}$ charge supply layer in between two undoped layers of AlGaN spacer and barrier [2]. The insersion of AlN spacer was also observed to improve the 2DEG density and the channel electron mobility, resulting in an improved $I_{ds, \max}$ of 1.2 A/mm [87].

In addition to I_{ds} , the transconductance (g_m) is another important parameter to indicate the response in I_{ds} to a change in the gate voltage (V_{gs}), which is defined as [88]:

$$g_m = \left. \frac{\partial I_{ds}}{\partial V_{gs}} \right|_{V_{gs}=\text{constant}} \quad \text{Eqn 2.1}$$

g_m is important for low-power consumption driving. J. W. Chung, et al. reported the maximum transconductance ($g_{m, \max}$) of 410 mS/mm using a T-gate ($L_g = 60$ nm) [3] D. Marti, et al. achieved an even higher $g_{m, \max}$ of 540 mS/mm with the source-drain space of 1 μm ($L_g = 100$ nm) [89].

One of the challenges in AlGaN/GaN HEMTs is the anomalous effects such as current collapse caused by the surface traps and/or traps in the heterostructures. The surface traps are typically due to the surface states exist at the surface of AlGaN/GaN if not properly passivated. The traps in the heterostructure are typically due to the non-optimal epitaxial growth. The current collapse is defined as the discrepancy between the DC and pulse I-V characteristics of AlGaN/GaN HEMTs, as shown in Figure 2-6. When device is pinched-off with high V_{DS} value (e.g. Q point in Figure 2-6), the presence of high electric at the edge of gate close to drain leads to an injection of electrons

from the gate metal to the surface traps [90][91]. Hence, the channel beneath will be depleted, as shown in Figure 2-7 (a). These trapped electrons will emit from the traps and increase the current in the channel when the device is switched on, as shown in Figure 2-7 (b). The slow electron emission will cause the slow recovery of current during pulse on-state transient hence lead to higher on-resistance and lower saturated current, as shown in Figure 2-6. The region with trapped electrons is referred as a virtual gate [92].

Figure 2-6 Typical pulsed I-V characteristics to show the current collapse of AlGaN/GaN HEMTs

Figure 2-6 also shows the linear fit for the extraction of static and dynamic on-resistance (static- and dynamic- R'_{on}) from the linear fit of pulsed I-V characteristics. The normalized static (or dynamic) on-resistance is given by:

$$R_{on} = R'_{on} \times A = R'_{on} \times W_g \times (L_{sg} + L_g + L_{gd}) \quad Eqn\ 2.2$$

The small value of R_{on} is preferred for low power-loss operation of

HEMT device. T. Ide, et al. has reported a R_{on} of $0.1 \text{ m}\Omega\cdot\text{cm}^2$ with a L_g of $0.1 \text{ }\mu\text{m}$ and a L_{sg} of $0.5 \text{ }\mu\text{m}$ [93]. To date, the lowest R_{on} ($0.09 \text{ m}\Omega\cdot\text{cm}^2$) was observed by M. Inada, et al. with a minimized source-drain space ($2.2 \text{ }\mu\text{m}$) but at a lower OFF-state breakdown voltage (35 V) [11].

Figure 2-7 (a) electron injection from gate and trapped at wafer surface at off-state (b) emission of trapped electron at on-state

2.4 Process flow of AlGaN/GaN HEMTs fabrication

The conventional process flow for the AlGaN/GaN HEMTs fabrication with 4-layer mask sets is shown in Figure 2-8

The whole process flow includes four steps namely, mesa isolation, ohmic contact formation, Schottky gate formation and surface passivation.

(1) Mesa isolation

In order to isolate adjacent devices, mesa isolation is needed. This is typically realized by dry etching, such as Reactive Ion Etching (RIE) or

Figure 2-8 Conventional process flow for AlGaIn/GaN HEMTs fabrication

Inductive Coupled Plasma-RIE (ICP-RIE). The mesa step starts from the coating of photoresist layer AZ5214 and the patterning using a UV-light (wave length: 405 nm) contact-mode mask aligner. After the development of photoresist by CD-26 developer, the photoresist residue is removed using O_2 plasma descum. The wafer is then loaded into AST Crie-200 chamber for dry etching. The etching is carried out with Cl_2/BCl_3 gas rate of 20/40 (unit: sccm), process chamber pressure of 12 mTorr and RF power of 100 W. The 180 s etching will create a mesa depth of ~ 150 nm, which is much deeper than the depth of buried

2DEG channel (~ 20 nm).

(2) Ohmic contact formation

Ohmic contact with low contact resistance is essential to achieve high drain current and low heat generation. The contact resistance of the ohmic contact can be influenced by a lot of factors, such as the structure and the quality of AlGaIn/GaN epi-layers, the doping concentration in AlGaIn barrier, the surface treatment prior to ohmic metal deposition, the ohmic metals and the thicknesses, the annealing temperature and duration, etc.

The ohmic contact step starts from the pad definition using the image reversal process of photoresist AZ5214. Firstly, a layer of AZ5214 with thickness ~ 1.5 μm is coated on the wafer. After the soft bake (hot plate 105 °C, 2 min.), the wafer is exposed by UV-light (wave length: 405 nm) using contact mode mask aligner (light intensity: 36 mW/cm², time: 1.1 s). Next, a hard bake (hot plate 105 °C, 2 min.) and the flood exposure are carried out (same UV-light intensity, 11 s). The ohmic pad patterns then are developed using CD-26 developer. After the O₂ descum and buffered oxide etchant (BOE) treatment for 2 min., Ti/Al/Ni/Au (20/120/40/50 nm) metal scheme is evaporated using an electron-beam (e-beam) system and then followed by metal lift-off using Acetone or AZ 300T photoresist stripper. Finally, the wafer is rapid thermal annealed (RTP) at 775 °C for 30 s at N₂ ambient. Figure 2-9 illustrates the steps of metal lift-off process.

Figure 2-9 Principles of metal lift-off process

(3) Schottky gate formation

Schottky gate formation is one of the most critical steps in AlGaIn/GaN HEMTs. The selected gate materials need to form a Schottky contact at the metal-semiconductor (M-S) interface. Typically, high work function metal Ni (50 nm) is used to provide the Schottky barrier height and Au cap layer (200 nm) is used to protect Ni from being oxidized. As the L_g of HEMT device discussed in this thesis is larger than 1 μm , contact mode lithography can be used to define the gate patterns. The fabrication of device with sub-micron L_g needs the electron beam lithography (EBL) for gate pattern writing.

The process procedure for gate formation is similar to that used for ohmic contacts. It should be pointed out that even a small alignment

error will cause short circuit in some devices or cause a malfunction of all test patterns in the whole die. Hence, the mis-alignment of gate mask with the alignment mark of ohmic contact layer should be strictly controlled to be smaller than 1 μm .

(4) Surface passivation

A large amount of surface states typically present at the surface of AlGaIn/GaN HEMT structure. The fabrication processes will also introduce more surface states due to the damage to the surface. These surface states which may consist of dangling bonds of the surface atoms, dislocations or defects generated during the epitaxial growth or during the processes, foreign contaminations, etc., will act as trap centers and result in degradation in device characteristics (e.g. current collapse in Section 2.4). The passivation step using the right dielectrics can effectively reduce the surface states and suppress the characteristics degradation of HEMT device. In this thesis, a 120 nm-thick Si_3N_4 layer deposited by plasma enhanced chemical vapor deposition (PECVD) at 300 °C is used as the passivation layer. The passivation is carried out in Cello Aegis-20 system with SiH_4/NH_3 gas rate of 55/145 (unit: sccm), chamber pressure of 400 mTorr, RF power of 50 W and process time 125 s. After the passivation, the contact pads are opened for probing purposes during measurement. The removal of the dielectric materials is achieved using CF_4 plasma etching with an RIE system.

2.5 Process modules for CMOS-compatible processes

Although conventionally III-V processes are used to produce excellent state-of-the-art AlGaIn/GaN HEMTs, they are not compatible with standard Si CMOS fabrication line. The main reasons include:

- The mesa isolation by plasma etching yields a non-planar surface on the wafer. Some issues are associated with this non-planar fabrication, e.g. the leakage current will increase due to the direct contact of gate metal and passivation layer with the 2DEG at the mesa side wall, etc.
- Ti/Al/Ni/Au ohmic contacts and Ni/Au Schottky gates have been used for AlGaIn/GaN HEMTs. However, gold is not allowed in Si CMOS line because of the contamination issue. In addition, Ni is not a preferred metal in Si CMOS line as it is difficult to be selectively removed by plasma etching.
- Metal lift-off process is typically used in III-V fabrication line for ohmic and gate contacts formation. However, in Si CMOS fabrication, the metal layer is first deposited on the entire wafer and then selectively removed by dry or wet etching.
- In Si CMOS fabrication, the gate electrode is deposited prior to the ohmic contact (source and drain). This is the so-called gate-first process. However, the conventional AlGaIn/GaN HEMT fabrication uses the gate-last process. To accommodate the gate-first process, the ohmic contact with low annealing temperature is needed. However, the

high annealing temperature ($\sim 800^{\circ}\text{C}$) in gold-based III-V processes renders the gate-first process unsuitable.

In order to realize a CMOS-compatible process for AlGaIn/GaN HEMTs, alternative approaches need be developed. In this thesis, three key process modules namely device isolation, ohmic contact and Schottky gate will be the focuses. For device isolation, instead of using dry etching which results in non-planarity, ion implantation will be used. For ohmic contact and Schottky gate, non-gold materials will be employed. The development of these process modules and the AlGaIn/GaN HEMT characteristics will be presented in details. In addition, the conduction mechanisms associated to the developed processes (e.g. ohmic contact and Schottky gate) will be systematically studied to better understand their impacts on the device characteristics.

Chapter 3 $^{131}\text{Xe}^+$ Implanted Isolation

Inter-device isolation is extremely important for the proper operation of functional integrated circuits (ICs) so that neighbouring devices are able to work independently without interfering with one another. Conventional mesa etching isolation in AlGaIn/GaN HEMTs fabrication will lead to a non-planar wafer surface, which is not compatible with standard Si CMOS process. This chapter starts with a review on the conventional mesa etching isolation and other reported isolation techniques. Next, the isolation effect using multi-energy ultra-heavy $^{131}\text{Xe}^+$ ion implantation and the investigation on the origins of high resistivity of implanted region will be discussed. Finally, the characteristics of AlGaIn/GaN HEMTs using $^{131}\text{Xe}^+$ implant isolation will be presented and compared with that using conventional mesa etching.

3.1 Conventional mesa etching isolation and other isolation techniques

Inter-device isolation of AlGaIn/GaN HEMTs is conventionally realized by mesa etching process using Chlorine-based plasma (BCl_3/Cl_2 , Cl_2/Ar , Cl_2 , etc.) by reactive ion etching (RIE) or inductively coupled plasma reactive ion etching (ICP-RIE) [1][59][68][94]. By selectively removing the AlGaIn/GaN HEMT structure and hence physically cutting off the 2DEG in between neighboring active regions, the inter-device isolation can be achieved, as shown

in Figure 3-1. The mesa etching should be much deeper than the location of 2DEG to ensure that there is sufficient isolation. Typically, the mesa etching depth of ~ 150 nm was used in the device structure used in this work.

Figure 3-1 Cross-sectional schematic diagram of AlGaIn/GaN HEMTs isolated by mesa etching

In order to improve the performance of mesa isolation, additional treatments were investigated. Z. Xu, et al. observed a reduced mesa current by three orders of magnitude after a post gate annealing (PGA) at 300 °C in N₂/O₂ ambient for 45 min. [95]. Similarly, the PGA at 350 °C in N₂/H₂ was also found to be helpful in suppressing the OFF-state source-drain current and the reverse gate leakage current [96]. Except for the annealing process, the passivation of surface states can also be done by other approaches. H. Kim, et al. proposed an in-situ O₂ plasma treatment after the mesa etching, by which the damaged surface was passivated and better pinch-off characteristics was observed [97].

X. Lu, et al. combined the surface treatment by O_2 plasma and HCl solution cycles with the PGA at 300 °C in N_2 ambient. An impressive surface leakage reduction by five orders of magnitude has been achieved [98]. Alternatively, T. Hashizume, et al. used a thin layer of Al followed by the ultra-high vacuum (UHV) annealing and the removal of Al layer. The suppression of surface leakage was attributed to the recovery of nitrogen-vacancy-related defects and the reduction in amount of oxygen impurities [99].

Except for the mesa etching isolation, other isolation technologies are reported for the fabrication of AlGaIn/GaN HEMTs. H. Masato, et al. tried to isolate the devices by thermal oxidization. A significantly reduced leakage current (24 μA to 65 pA) was observed with oxidized Ga_2O_3 and the 2DEG density as well as the carrier mobility were not compromised [100]. Y. -S. Lin, et al. reported the use of refilled SiO_2 in the etched region to form the isolation [96]. Compared to the conventional mesa etching, both the OFF-state source-drain current and gate leakage of AlGaIn/GaN HEMTs were reduced by one order of magnitude using the oxide refill technology. The fluorine-based plasma treatment was also reported to be able to deplete the 2DEG and hence form an isolation region with planar surface [101][102]. The leakage current of isolation region using fluorine-based plasma treatment was at the same level as that using mesa etching [101]. In addition to the dry etching by plasma, mesa isolation can also be achieved using UV-photo assisted wet etching in KOH solution as reported by H. Maher, et al [103][104]. Pt metal layer and $K_2S_2O_8$ were used as the hard mask and the oxidizer, respectively. The relatively low etch rate (~ 2

nm/min) makes the method more suitable for the formation of shallow mesa or gate recess.

3.2 Isolation by Ion-Implantation

Although the mesa etching process has been successfully adopted for the inter-device isolation, the disadvantages of this technology hinder its further implementation in CMOS-compatible fabrication of AlGaIn/GaN HEMTs. Due to the presence of a mesa sidewall in the conventional mesa-isolation process, the metal gate (as well as the passivation layer) is in contact with the 2DEG of AlGaIn/GaN HEMT structure, as illustrated in Figure 3-2. This will lead to higher gate leakage current and non-uniformity in buffer breakdown voltage [105][106]. Moreover, the AlGaIn/GaN wafer after mesa etching isolation exhibits a non-planar surface, which in return affects the process yield [63][65]. Finally, the three-dimensional mesas may impose significant limitations on the minimum size realized by photolithography and are not desirable for digital IC applications that require high-density and high-uniformity circuit integration.

In order to avoid the issues introduced by plasma etching, isolation by ion-implantation using various ion species, such as N^+ [32][65], P^+/He^+ [66], O^+ [63][67], Fe^+ [64], Ar^+ [106][107][108], Zn^+ [109] and Kr^+ [106][107], have been studied to realize planar AlGaIn/GaN HEMTs, as summarized in Table 3-1. However, it was observed that most of the ion species are not able to maintain the high resistivity of the implanted region after high temperature

annealing except for Fe^+ [64] and Kr^+ [106][107]. The sheet resistance (R_{sh}) of P^+/He^+ implanted region was even reported to decrease by ~ 6 orders of magnitude after 900 °C annealing [66].

Figure 3-2 Gate leakage from the mesa sidewall due to the direct contact of gate metal with 2DEG [110]

Table 3-1 Summary of isolation implant with various ion species, energies, and doses for AlGaN/GaN HEMTs fabrication before thermal treatment

Ions	Al (%)	Dose ($\times 10^{13} \text{ cm}^{-2}$)	Energy (keV)	R_{sh} ($\Omega/\square$)	Ref
N^+	0.25	0.6/1.8/2.5	30/160/400	3×10^{11}	[65]
P^+/He^+	0.15	(P)0.05/0.19/(He)6	75/180/75	7×10^{11}	[66]
O^+	0.3	50/50/50	25/50/75	10^{12}	[63][67]
Fe^+	0.25	3	50	5×10^{12}	[64]
Ar^+		4	40	4×10^{12}	
B^+		4.5	10	5×10^{12}	
N^+		5	15	5×10^{12}	
Zn^+	0.3	1.9/3.8/3.5/13	30/80/175/350	5.9×10^{11}	[109]
N^+	0.15	20/65	25/120	-	[32]
Ar^+	0.26	0.42/0.63/1.11	5/15/45	-	[106]
Kr^+		0.74/0.71/1.11	5/25/45	-	[107]
Ar^+	0.15	34	160	-	[108]

Figure 3-3 shows the transport of ions in matter (TRIM) simulated displacements in AlGaIn/GaN HEMT structure by ion implantation at 50 keV using $^{14}\text{N}^+$, $^{40}\text{Ar}^+$, $^{56}\text{Fe}^+$, $^{84}\text{Kr}^+$ and $^{131}\text{Xe}^+$. With a heavier ion mass, the displacements were at the shallower location with higher concentration. Owing to the 2.33 times and 1.56 times heavier mass as compared to $^{56}\text{Fe}^+$ and $^{84}\text{Kr}^+$ respectively, $^{131}\text{Xe}^+$ is expected to introduce more lattice damage and disorders and thus is able to withstand subsequent high temperature processes.

Figure 3-3 TRIM calculation of $^{14}\text{N}^+$, $^{40}\text{Ar}^+$, $^{56}\text{Fe}^+$, $^{84}\text{Kr}^+$ and $^{131}\text{Xe}^+$ -ions profile in AlGaIn/GaN HEMT structure

By far, the implant-isolation of AlGaIn/GaN HEMTs device by ultra-heavy $^{131}\text{Xe}^+$ ions has not been reported. This work systematically investigates the inter-device isolation of multi-energy $^{131}\text{Xe}^+$ implantation in terms of R_{sh} , ON/OFF current ratio ($I_{\text{on}}/I_{\text{off}}$), buffer blocking voltage (BV_{buff}), OFF-state breakdown voltage (BV_{gd}) and DC output and transfer characteristics. The

comparison to the conventional mesa-isolation shows that the $^{131}\text{Xe}^+$ implantation gives better inter-device isolation with improved leakage and breakdown characteristics.

3.3 Experimental details

The AlGaIn/GaN HEMT structure was grown on 4-inch diameter high-resistivity Si (111) substrate by Metal-organic chemical vapor deposition (MOCVD). At room temperature, the 2DEG sheet carrier density, the channel mobility, and R_{sh} are measured as $\sim 9.7 \times 10^{12} \text{ cm}^{-2}$, $\sim 1450 \text{ cm}^2/\text{V}\cdot\text{s}$, and $440 \text{ }\Omega/\square$, respectively. After a blanket coating of 120 nm-thick Si_3N_4 by plasma-enhanced chemical vapor deposition (PECVD) at 300°C , a selective etching of Si_3N_4 by CF_4 -based plasma was performed to expose the region for isolation. Next, the sample was subject to $^{131}\text{Xe}^+$ implantation which was carried out at the GNS Science in New Zealand.

The implantation uses multi-energy of 5/15/30/45/100 keV with respective doses of $5 \times 10^{13}/1 \times 10^{13}/1 \times 10^{13}/1 \times 10^{13}/1 \times 10^{13} \text{ cm}^{-2}$ to ensure the uniform damages up to the GaN buffer layer [111]. The sample surface was 3° tilted to the incident $^{131}\text{Xe}^+$ ion beam during the implantation to avoid the ion channeling effect. Figure 3-4 gives the key components as well as the photos of the ion implanter used for this study.

Figure 3-4 Key components and the photo of the ion implanter used for $^{131}\text{Xe}^+$ implantation isolation (Ion-implanter at GNS Science, New Zealand)

TRIM simulation indicates that $^{131}\text{Xe}^+$ ions are able to penetrate up to the depth of ~ 80 nm, as shown in Figure 3-5 (a). The vacancy concentrations in Figure 3-5 (b) suggest that except for the energy of 5 keV with the dose of $5 \times 10^{13} \text{ cm}^{-2}$, the rest of the implantations create a relative uniform maximum vacancy density of $\sim 1.3 - 2 \times 10^{20} \text{ cm}^{-3}$. Followed by the removal of Si_3N_4 hard mask in buffered oxide etchant (BOE) for 10 min. to expose the active region, ohmic contacts with Ti/Al/Ni/Au (20/120/40/50 nm) stacks were deposited by e-beam evaporator. The contact resistance (R_c) after rapid thermal annealing (RTA) at 775 °C for 30 s in N_2 ambient was measured to be $\sim 0.3 - 0.4 \Omega \cdot \text{mm}$. The Ni/Au (50/200 nm) gate was then defined by e-beam evaporation and lift-off process. Finally, the sample was passivated with 120 nm-thick Si_3N_4 by

PECVD at 300 °C.

Figure 3-5 Depth profiles of (a) $^{131}\text{Xe}^+$ ions and (b) vacancy concentrations at various energies and doses by TRIM simulation.

For comparison, the devices were also fabricated using conventional mesa-isolation by BCl_3/Cl_2 plasma etching on the same AlGaIn/GaN HEMT structure [68][112]. The mesa etching targets to a depth of 150 nm. The actual etching depth measured by Dektak XT surface profiler is ~ 165 nm with good uniformity, as shown in Figure 3-6, which is about two times of the maximum implantation depth of $^{131}\text{Xe}^+$ ions (See Figure 3-5). Figure 3-7 gives the three-dimensional schematic diagram of the fabricated AlGaIn/GaN HEMTs by (a) $^{131}\text{Xe}^+$ implant-isolation and (b) conventional mesa-isolation. Figure 3-8 shows the optical microscope images (top view) of the ohmic contacts and the conductive channel with (a) $^{131}\text{Xe}^+$ implant-isolation and (b) mesa-isolation. It is clear that the implant-isolated sample does not have the edge-defined mesas when compared with the mesa-isolated sample. Since the mesa channel is not

formed in the implant-isolated device, the gate metal will not touch the sidewall of the 2DEG channel (see Figure 3-7 (a)).

Figure 3-6 The Surface profile of AlGaIn/GaN HEMT structure isolated by mesa etching. The etching depth is ~ 165 nm.

Figure 3-7 Three-dimensional schematic diagrams of AlGaIn/GaN HEMTs isolated by (a) $^{131}\text{Xe}^+$ implantation and (b) Mesa etching. (The Si_3N_4 passivation layer is not included)

Figure 3-8 Optical microscope images (100 $\times$) of 5 μm conductive channel with
(a) $^{131}\text{Xe}^+$ implant-isolation and (b) mesa-isolation.

The fabricated AlGaN/GaN HEMTs and buffer/channel patterns were characterized using Agilent B1505A power device analyzer. The buffer/channel current-voltage characteristics ($I_{\text{on}}/I_{\text{off}}$ ratio, R_{sh} , and activation energy) and BV_{buff} of buffer were measured on 5 μm -gap and 10 μm -gap test structures, respectively, with a same width of 50 μm . The DC output and transfer characteristics were obtained from single-gate HEMT with the source-gate distance (L_{gs}) and the gate length (L_{g}) of 2 μm , and the gate-drain distance (L_{gd}) of 4 μm with a gate width (W_{g}) of 80 μm . The three-terminal OFF-state breakdown (BV_{gd}) characterization was performed by varying L_{gd} from 4 μm to 10 μm with the same L_{g} , L_{gs} and W_{g} .

3.4 Results and Discussions

Figure 3-9 (a) and (b) show the I-V characteristics of channel and buffer

structures isolated by $^{131}\text{Xe}^+$ implantation and conventional mesa etching. The measured channel current density is $\sim 780 \text{ mA/mm}$ for both implant- and mesa-isolated channel, which is consistent to the measured R_{sh} of $444.25 \pm 8.96 \text{ } \Omega/\square$ for implant-isolated and $458.72 \pm 1.54 \text{ } \Omega/\square$ for mesa-isolated sample, suggesting that the 2DEG concentration and the carrier mobility in the active region are not affected by $^{131}\text{Xe}^+$ implantation.

Figure 3-9 Buffer and channel I-V characteristics of (a) $^{131}\text{Xe}^+$ implant-isolated and (b) mesa-isolated sample, and statistical analysis of measured (c) $I_{\text{on}}/I_{\text{off}}$ ratio and (b) R_{sh} for both implant-isolated and mesa-isolated samples.

However, the implant-isolated sample exhibits one order of magnitude

lower buffer leakage current, which is mainly caused by large lattice damage and disorders created by the ultra-heavy $^{131}\text{Xe}^+$ ions [107][113]. The $I_{\text{on}}/I_{\text{off}}$ ratio, which is defined as the ratio of channel current and buffer leakage at the bias of 20 V, are $\sim 1.4 \times 10^8$ and $\sim 2.7 \times 10^8$ for implant- and mesa-isolated sample, respectively. The statistical $I_{\text{on}}/I_{\text{off}}$ ratio and R_{sh} measured from 10 different locations on each sample are plotted in Figure 3-9 (c) and (d) with summarized mean and standard deviation (SD) values given in Table 3-2. The observation of small variations in $I_{\text{on}}/I_{\text{off}}$ ratio and R_{sh} indicates the uniform inter-device isolation for both isolation techniques. The one order of magnitude higher R_{sh} in implant-isolated sample is in good agreement with the measured $I_{\text{on}}/I_{\text{off}}$ ratio. The improvement in $I_{\text{on}}/I_{\text{off}}$ and R_{sh} are solely due to the reduction of buffer leakage by multi-energy ultra-heavy $^{131}\text{Xe}^+$ ion-implantation.

For a better understanding on the origins of isolation by $^{131}\text{Xe}^+$ implantation, the statistical R_{sh} at various temperatures are plotted in Figure 3-10 to extract the activation energy (E_A) of isolated AlGaIn/GaN HEMT structure. The measurement was carried out from 300 to 420 K, with a step increment of 30 K. For the high-resistive isolated region, it is difficult to extract the carrier concentration and the mobility by Hall measurement. Hence, R_{sh} is used for E_A extraction assuming that the dependency of mobility on the implantation conditions is negligible as reported in [64]. From the slope of the linear fit in Figure 3-10, the extracted E_A is 0.513 eV.

H. Umeda, et al. reported the extraction of E_A from Ar^+ and Fe^+ implant-isolated AlGaIn/GaN HEMTs. For Fe^+ implant-isolation, E_A was reduced from

~ 0.7 eV to ~ 0.35 eV after 20 min. annealing at and above 800 °C. They claimed that $E_A \sim 0.7$ eV was related to lattice damage by ions whereas the reduced $E_A \sim 0.35$ eV was associated to the deep level of Fe^+ ions at Ga sites. On the other hand, although R_{sh} of Ar^+ implant-isolation rapidly decreased after high temperature (> 800 °C) annealing, E_A at ~ 0.5 eV was nearly not influenced by annealing up to 1200 °C, suggesting that the origins of high resistivity were solely due to the recoverable lattice damage [64]. Similar to the case of Ar^+ implant-isolation, $^{131}\text{Xe}^+$ ions cannot occupy Ga sites and then form deep levels in the lattice as both Ar^+ and $^{131}\text{Xe}^+$ are inert gas species. Therefore, the extracted $E_A \sim 0.513$ eV denotes the energy level of lattice damage by $^{131}\text{Xe}^+$ implantation, which is also the origin of the high-resistivity of implanted region.

Figure 3-10 Arrhenius plot on statistical value of R_{sh} at the temperature range of 300-420 K

The breakdown characteristics of implant- and mesa-isolated buffers with and without Si_3N_4 passivation were measured by fixing a current compliance of 0.5 mA/mm. As shown in Figure 3-11 (a) and (c), the buffer breakdown voltages (BV_{buff}) of un-passivated device are 447 and 419 V for implant- and mesa-isolation, respectively. The breakdown voltages are mainly limited by the 2.4 μm i-GaN buffer and transition layer (See Figure 3-7) [114].

Figure 3-11 Buffer breakdown characteristics and statistical BV_{buff} of (a), (b) $^{131}\text{Xe}^+$ implant- and (c), (d) mesa-isolated AlGaIn/GaN HEMT structure with and without Si_3N_4 .

Compared to the sharp decrease of BV_{buff} (221 V) in Figure 3-11 (c) in

mesa-isolated buffer, BV_{buff} of implant-isolated buffer in Figure 3-11 (a) is almost unchanged (443 V) with Si_3N_4 passivation. The BV_{buff} values measured from 10 different locations on each sample are plotted in Figure 3-11 (b) and (d) and are tabulated in Table 3-2. The uniform BV_{buff} in implant-isolated buffer with negligible change before and after passivation (see Figure 3-11 (b)) is also observed in AlGaIn/GaN HEMT structures isolated by Ar^+ and Kr^+ implantation [107]. On the contrary, BV_{buff} with large variations ranging from 152 to 436 V (SD: ± 99.8 V, see Table 3-2) with Si_3N_4 passivation are observed in mesa-isolated buffer in spite of the uniform BV_{buff} without passivation (see Figure 3-

Table 3-2 Inter-device isolation and device parameters of AlGaIn/GaN HEMTs fabricated with $^{131}\text{Xe}^+$ implant-isolation and mesa-isolation (Statistical Analysis was carried out on 10 different patterns with the same dimensions).

Meas. On.	Parameters	$^{131}\text{Xe}^+$	Mesa
Buffer/Channel	$R_{\text{sh}} (\times 10^9 \Omega/\square)$	9.44 $\pm$ 4.91	1.10 $\pm$ 0.59
	$I_{\text{on}}/I_{\text{off}} (\times 10^7)$	19.05 $\pm$ 9.18	2.15 $\pm$ 1.07
	BV_{buff} w/o pas- Si_3N_4 (V)	442.9 $\pm$ 6.9	439.3 $\pm$ 8.6
	BV_{buff} with pas- Si_3N_4 (V)	410.6 $\pm$ 10.2	325.3 $\pm$ 99.8
AlGaIn/GaN HEMT devices	$I_{\text{ds, max}} @ V_g = 1 \text{ V}$ (mA/mm)	596.9	592.9
	V_{th} (V)	-3.57	-3.6
	Subthreshold Swing (mV/dec)	111.4	127.8
	$g_{\text{m, max}}$ (mS/mm)	155.1	157
	$I_{\text{ds, OFF}} @ V_g = -5 \text{ V}$ ($\mu\text{A}/\text{mm}$)	1.9	13.1
	$I_g @ V_g = -5 \text{ V}$ ($\mu\text{A}/\text{mm}$)	2.4	20.1

11 (d)). The different effects of Si_3N_4 passivation on BV_{buff} can be explained by the mesa sidewall leakage. For mesa-isolated AlGaIn/GaN HEMT structure, the coverage of Si_3N_4 on the mesa sidewall may have introduced additional leakage paths due to the direct contact of 2DEG with Si_3N_4 layer [110]. Surface treatment before the Si_3N_4 deposition may help to reduce the additional leakage path [115].

Figure 3-12 shows (a) DC output and (b) DC transfer characteristics of AlGaIn/GaN HEMTs isolated by $^{131}\text{Xe}^+$ implantation and by mesa etching with Si_3N_4 passivation. The key parameters of AlGaIn/GaN HEMT devices are summarized in Table 3-2. The implant-isolated device exhibits very close values in maximum source-drain current ($I_{\text{ds, max}}$), threshold voltage (V_{th}), subthreshold swing (SS), and maximum transconductance ($g_{\text{m, max}}$) as compared

Figure 3-12 (a) DC output and (b) DC transfer characteristics of AlGaIn/GaN HEMTs with $^{131}\text{Xe}^+$ implant-isolation and mesa-isolation

to the mesa-isolated device, indicating the DC performance of HEMT is not compromised by $^{131}\text{Xe}^+$ implant-isolation. As shown in Table 3-2, both the OFF-state source-drain leakage ($I_{\text{ds, OFF}}$) and the gate leakage (I_{g}) are reduced by one order of magnitude by implant-isolation, which is in good agreement with the observation of large $I_{\text{on}}/I_{\text{off}}$ ratio and R_{sh} in Figure 3-9.

Figure 3-13 shows the three terminal OFF-state breakdown (BV_{gd}) characteristics of (a) $^{131}\text{Xe}^+$ implant-isolated and (b) mesa-isolated AlGaIn/GaN HEMTs with varying L_{gd} at $V_{\text{g}} = -5$ V. BV_{gd} is defined as the voltage at which I_{ds} reaches the compliance of 0.1 mA/mm. As shown in Figure 3-13 (c), BV_{gd} of implant-isolated HEMTs was increased by $\sim 106 - 128$ V as compare to that of mesa-isolated devices with the same L_{gd} . The improvement of BV_{gd} in implant-isolated AlGaIn/GaN HEMT is consistent with the decrease of reverse gate leakage (I_{gd}) current, which is one order of magnitude lower than the HEMTs with mesa-isolation (See Figure 3-13 (d)). The error bar at $V_{\text{R}} = -10$ V and $V_{\text{R}} = -20$ V are based on statistical value of 5 different HEMTs ($L_{\text{gd}} = 4 \mu\text{m}$) measured on each sample. The observation of enhanced BV_{gd} from implant-isolated HEMT is due to the low gate leakage current. Similar observation has also been realized in AlGaIn/GaN HEMTs by Ar^+ and Kr^+ implantation [106]. P. Liu, et al. observed an elimination of gate leakage by placing plasma enhanced tetraethyl-orthosilicate (PE-TEOS) oxide spacers in between gate and mesa sidewalls [116]. Compared to the insertion of oxide spacer, our implant-isolation by $^{131}\text{Xe}^+$ ions provides a more practical way to avoid the direct contact of gate with 2DEG at the mesa sidewalls which could significantly

reduce the reverse gate leakage.

Figure 3-13 Three terminal OFF-state breakdown characteristics of (a) mesa- and (b) implant-isolated AlGaIn/GaN HEMTs; (c) BV_{gd} vs. L_{gd} and (d) two terminal gate-leakage current of both mesa- and implant-isolated AlGaIn/GaN HEMTs

3.5 Summary

In summary, multi-energy implantation by ultra-heavy inert gas $^{131}\text{Xe}^+$

was implemented on AlGaIn/GaN HEMTs for device isolation. This technique offers improved inter-device isolation with one order of magnitude lower buffer leakage current and one order of magnitude higher for both ON/OFF current ratio and sheet resistance than the conventional mesa-isolation. The extracted activation energy of implant-isolation by Arrhenius plot is 0.513 eV which denotes the energy levels of lattice damage by $^{131}\text{Xe}^+$ ions. In addition to the stable buffer breakdown voltage after Si_3N_4 passivation, implant-isolated AlGaIn/GaN HEMTs exhibit improved OFF-state breakdown voltage by $\sim 106 - 128$ V, one order lower reverse gate leakage without compromising on the DC output and transfer characteristics as compared to the mesa-isolated device. The improvement in higher OFF-state breakdown voltage, lower gate leakage and more stable buffer breakdown voltage after passivation is due to the elimination of direct contact of deposited Ni/Au gate and Si_3N_4 passivation with 2DEG at the mesa sidewalls. These results indicate that the ultra-heavy $^{131}\text{Xe}^+$ ions are promising for the development of planar AlGaIn/GaN HEMT technology with improved breakdown and leakage characteristics.

Chapter 4 Non-gold Ta/Si/Ti/Al/Ni/Ta Ohmic

Contact

Ohmic contact is defined as a Metal-Semiconductor (M-S) contact that has a negligible contact resistance relative to the bulk or spreading resistance of the semiconductor [117]. Conventional ohmic contacts used for AlGaN/GaN HEMTs typically employ gold-based metal stack such as Ti/Al/X/Au, where X denotes barrier metal such as Ni, Ti, Mo, etc. As the usage of gold will cause contamination in Si CMOS processing line, the development of non-gold ohmic contact is essential to realize the CMOS-compatible processes for AlGaN/GaN HEMTs fabrication. In this chapter, the basic physics of an ohmic contact will be briefly discussed in the first section. Next, the Transmission Line Model (TLM) measurement technique will be introduced. A brief summary of the gold-based and non-gold ohmic contacts will then be given in the next two sections. The characterization on non-gold Ta/Si/Ti/Al/Ni/Ta ohmic contact and the investigation on the reasons for the achieved low contact resistance (R_c) value are then discussed in the next section. Finally, the conduction mechanism of our non-gold Ta/Si/Ti/Al/Ni/Ta ohmic contacts is systematically studied and presented.

4.1 Electron Transport across a Metal-Semiconductor Interface

Interface

The band diagrams of a metal with an n-type lightly, moderately and heavily doped semiconductors are given in Figure 4-1 (a), (b) and (c), respectively. The M-S interface here universally represents the contact system that consists of a metal and a semiconductor material. From Figure 4-1, the barrier heights at the semiconductor side are different ($q\phi_L < q\phi_M < q\phi_H$) due to the different doping level. Thus, the electron transports through these interfaces are dominated by different mechanisms.

Figure 4-1 Electron transport through the M-S interface

In Figure 4-1 (a), the lightly doped semiconductor yields a wide depletion region. Due to the relatively smaller ϕ_L and larger depletion width (W_L), the electrons are more likely to surmount the barrier with sufficient thermal energy

than to tunnel through the barrier. This mechanism is known as thermionic emission (TE). When the doping level increases, the Schottky barrier becomes higher and the depletion width is reduced accordingly, as shown in Figure 4-1 (b). The electrons are able to tunnel through as well as to surmount the barrier. The total current across the M-S interface is contributed by the combination of thermionic emission and field emission, which is also known as thermionic-field emission (TFE). With further increasing of doping level, the Schottky barrier at the interface becomes very thin (even to atomic level) (see Figure 4-1 (c)). The electrons with high kinetic energy could penetrate through rather than surmount the barrier. This is so-called the field emission (FE) or tunneling.

The parameter $k_B T/E_{00}$ is used to estimate the tunneling possibility of electrons through a M-S interface. E_{00} is given by,

$$E_{00} = \frac{qh}{4\pi} \sqrt{\frac{N_D}{m^* \cdot \varepsilon}} \quad Eqn\ 4.1$$

where k_B , T , q , h , N_D , m^* and ε denote the Boltzmann constant, the temperature, the electron charge, Planck constant, the doping concentration, the effective electron mass and the dielectric constant of the semiconductor, respectively. When E_{00} is larger than the thermal energy ($k_B T$), the probability of the electron transport by tunneling will increase. For lightly doped semiconductor where $k_B T/E_{00} \gg 1$, thermionic emission dominates in the overall current. Conversely, a heavy doping results in $k_B T/E_{00} \ll 1$, suggesting an overall current that is dominated by tunneling. For other cases ($k_B T/E_{00} \approx 1$), both mechanisms contribute simultaneously [117].

The specific contact resistivity (ρ_c) of an M-S interface equals to the reciprocal value of the derivative of current density with respect to voltage at zero bias. The current-voltage relations have been developed based on the simple energy band models in Figure 4-1 using the Wentzel-Kramers-Brillouin (WKB) approximation [118], [119],

For TE mechanism,

$$\rho_c = \frac{k_B}{qA^*T} \exp\left(\frac{q\Phi_B}{k_B T}\right) \quad Eqn 4.2$$

For FE mechanism,

$$\rho_c = C_1 \exp\left(\frac{q\Phi_B}{E_{00}}\right) \quad Eqn 4.3$$

where C_1 is a factor with weak temperature dependence

For TFE mechanism,

$$\rho_c = C_2 \exp\left[\frac{q\Phi_B}{\sqrt{N_D} \cdot \coth(E_{00}/k_B T)}\right] \quad Eqn 4.4$$

where C_2 is a function of $q\Phi_B$ and T .

From *Eqns 4.2 to 4.4*, two effective methods can be used to reduce the ρ_c value of an M-S contact, which are (i) to decrease the Schottky barrier height (minimize $q\Phi_B$), and/or (ii) to increase the doping concentration (maximize N_D) at the vicinity region of interface irrespective of the kind of electron conduction that dominants in the total current.

4.2 Transmission Line Model (TLM) Measurement

From the device application point of view, the ohmic contact is expected to have very low resistance so as to consume minimal power and generate little heat. However, in practical cases, there always exists a finite value of contact resistance which impedes the flow of current. Typically, the contact resistance (R_c) and the specific contact resistivity (ρ_c) are the two key parameters used to evaluate the electrical performance of the ohmic contact. The transmission line model (LTLM) measurements, including linear-TLM (LTLM) and circular-TLM (CTLM), are commonly used to extract the value of R_c and ρ_c .

Figure 4-2 shows a LTLM pattern in a heterostructure. The transfer length (L_T) is defined as the distance from the edge that the current falls to $1/e$ of its original value. When a bias is applied on two contact pads, a current flows between two ohmic pads along the dashed lines are shown in Figure 4-2.

Figure 4-2 A linear transmission line model of current flow through two contacts in a heterostructure

The equivalent circuit of this model can be simplified as three resistors in series connection. Thus, the total resistance between two pads is given by [120]

$$R_{total} = 2R'_c + R_{sh} \frac{d}{W} \quad Eqn 4.5$$

where R_{sh} is the sheet resistance of semiconductor [unit: $\Omega/\square$ (or Ω/sq)], and R'_c is the contact resistance of an M-S interface (unit: Ω).

when R'_c is normalized by the width (W), *Eqn 4.5* is rewritten as

$$R_{total} = 2 \frac{R_c}{W} + R_{sh} \frac{d}{W} \quad Eqn 4.6$$

The specific contact resistivity (ρ_c) is defined as

$$\rho_c = \frac{R_c^2}{R_{sh}} \quad Eqn 4.7$$

The transfer length (L_T) can be obtained by the relationship:

$$L_T = \sqrt{\frac{\rho_c}{R_{sh}}} \quad Eqn 4.8$$

It should be pointed out that *Eqn 4.6* is only applicable to LTLM pattern. The circular-TLM is also widely used for ohmic resistance evaluation [121]. Figure 4-3 gives the layout of both LTLM and CTLM. With the implementation of CTLM patterns, the mesa recess process can be skipped and the current crowding effect at edge of pad is negligible. However, the correction factors are needed for the linear fit of the CTLM, which makes the calculation more complicated [9][10].

In this study, two sets of LTLM patterns are used for the calculation of R_c and ρ_c . The gap lengths of two LTLM sets are 5, 10, 20, 40, 80 μm and 3, 6,

9, 12, 15, 18 μm , respectively, as shown in the microscope image in Figure 4-4. The typical R_{total} versus d characteristics of LTLM structure is shown in Figure 4-5. The linear fitted straight line intercepts the R_{total} - axis at $2R_c/W$, and further intercepts d axis at $-2L_T$. As both W and d are design constants, R_c and ρ_c hence can be straightforwardly obtained.

Figure 4-3 Patterns of (a) Linear and (b) Circular TLM [121]

Figure 4-4 The microscope image of LTLM patterns used in this study

Figure 4-5 Linear fit of measured points of a LTLM structure

4.3 Conventional Ti/Al/X/Au Ohmic Contacts

In order to achieve very low ρ_c , a small Schottky barrier height ($q\Phi_B$) and/or a heavily doped region at the M-S interface are required based on Eqn 4.2 - 4.4. The localized heavily doping can be realized by the incorporation of dopant species (Si typically) into the (Al)GaN by implantation [124] or by diffusion [125]. For doping by ion implantation, a high-temperature annealing process (1500 °C in Ref [124]) is typically needed to activate the dopant atoms, which is not preferred in device fabrication. The thermal treatment with more time is also required to facilitate the dopant diffusion and activation in the diffusion method [125]. W. L. Lai, et al. reported that only 4.5% of implanted Si dose can be activated with 1 minute annealing at 750 °C, and 20% activation efficiency had been achieved at 1000 °C in p-GaN [126]. In the AlGaN thin

film, the activation efficiency is even lower due to the higher bonding energy of Al-N compared to that of Ga-N [127]. Briefly, the fabrication of low-resistance ohmic contact solely by the localized heavily doping is not easy in AlGaN/GaN HEMTs.

Hence, the types of materials to yield low $q\Phi_B$ need to be carefully chosen in the formation of ohmic contacts for AlGaN/GaN HEMTs. When the Fermi level pinning (FLP) effect is neglected, the magnitude of $q\Phi_B$ is directly determined by the Schottky-Mott theory [128], given by,

$$q\Phi_B = q\Phi_m - \chi_s \quad \text{Eqn 4.9}$$

in which $q\Phi_m$ and χ_s denote the metal work function and the electron affinity of semiconductor, respectively. Apparently, a small $q\Phi_B$ can be obtained only if the values of $q\Phi_m$ and χ_s are well matched. Given by $\chi_s(\text{GaN}) = 4.1 \text{ eV}$ [129] and $\chi_s(\text{Al}_x\text{Ga}_{1-x}\text{N})$ decreases with increasing Al mole fraction (x) [130], the materials with small $q\Phi_m$ is preferred in the formation of ohmic contacts onto AlGaN/GaN HEMTs.

Very few metals have $q\Phi_m$ value smaller than 4.1 eV, except for Mg [131][132]. Furthermore, even if one uses the metal that has a $q\Phi_m$ value close (or even smaller) to $\chi_s(\text{GaN})$, such as Mg ($\sim 3.66 \text{ eV}$) [131][132], Ag ($\sim 4.26 \text{ eV}$) [133], Al ($\sim 4.28 \text{ eV}$) [131][132], and Ti ($\sim 4.33 \text{ eV}$) [132], etc., the measured Schottky barrier height is not as small as the theoretical value calculated by Eqn 4.9. The FLP effect is widely observed in different metals on GaN surface [134][135]. In addition to metal work function, some other factors also need to be considered in ohmic contact design. For example, the selected

metal should exhibit low resistivity and comparable thermal expansion coefficient to (Al)GaN. Equally important, the first layer of a ohmic stack must have a good adhesion to (Al)GaN surface and hence to ensure no delamination of metal stack during the lift-off process. Based on the factors discussed above, the formation of good ohmic contacts on AlGaN/GaN HEMTs is a challenging work to GaN device researchers.

The conventional gold-based ohmic contacts used for AlGaN/GaN HEMTs typically include four metal layers, which are Ti, Al, barrier metal X, and gold from bottom to top. The barrier metal X is primarily used to confine the downward diffusion of gold. If gold is in direct contact with the AlGaN surface, the Schottky barrier at the M-S will be considerably high due to the large work function of gold (~ 5.1 eV [136]). For the barrier metal X, many different types of materials have been explored by various research groups. Metals such as Pt [137] [138] [139], Ni [137] [138] [139][68][140], Mo [137] [139], Ti [137] [139], Cu [141], Nb [137] [139], Ir [137] [139], Ta [137] [139], etc., have been assessed as the barrier metal X in ohmic contacts for AlGaN/GaN HEMTs. The top layer of gold is required to protect the rest of the metals beneath from being oxidized in the air. Moreover, the extremely low electrical resistivity of gold has an additional advantage in forming the ohmic contact with low resistance. The key component of the Ti/Al/X/Au ohmic stack is the Ti/Al bi-layer. The selection of Ti as the first metal layer is based on two important reasons. First, the work function of Ti is relatively low (~ 4.33 eV [17][18]). Second and more importantly, Ti is able to react with (Al)GaN

through a replacement chemical reaction as below:

The reaction in *Eqn 4.10* is driven by the formation of enthalpy differences between different nitrides. As reported by B. Van Daele, et al. the formation enthalpies of GaN, TiN and AlN are -110.9 , -265.5 and -318.1 kJ/mol, respectively, suggesting AlN as the most stable compounds among these three nitrides while GaN as the least stable one [144]. If Ti is deposited in contact with (Al)GaN followed by an annealing process to supply sufficient thermal energy, the reaction described in *Eqn 4.10* occurs.

The reaction of Ti and (Al)GaN produces two outcomes which are helpful in the formation of low resistance ohmic contact. Firstly, the out-diffusion of N-atoms will leave large amount of N-vacancies in the (Al)GaN [145]. As a kind of n-type dopant, the N-vacancies are able to enhance the electron transport across the M-S interface by tunneling mechanism. Compare to the previously discussed implantation and the diffusion methods [124][125], the chemical reaction is a more efficient method to create a localized heavily doping area with high carrier concentration. Secondly, as the key product of the interface reaction, the semi-metallic TiN exhibits a lower metal function ~ 3.74 eV as well as a higher electrical conductivity than that of Ti [145]. The direct contact of reacted TiN to the surface of AlGaN/GaN will result in a small Schottky barrier due to the reduced work function of TiN as compared to Ti. Furthermore, the semi-metallic TiN in pocket shape, in most cases, is observed to diffuse downward along the lattice dislocation and to penetrate through the

AlGa_N barrier and finally access the Ga_N buffer, as shown in Figure 4-6 [69][146][145]. Obviously, the direct link of annealed metal alloy through the conductive TiN pockets to 2DEG greatly facilitates the electron transport across the M-S interface, and hence reduces the contact resistance. In addition, Ti is also reported to be able to clean the native oxide on the surface of AlGa_N during the reaction [147], which is an extra advantage in the formation of low-resistance ohmic contact.

Figure 4-6 Cross-sectional TEM images of (a) Ti/Al/Ni/Au [69] (b) Ti/Al/Ti/Au [146] (c) Ti/Al/Mo/Au [145] ohmic contacts on AlGa_N/Ga_N HEMT structure.

The Al layer is primarily used to consume the excessive Ti material in the reaction given by Eqn 4.10 which will damage the AlGa_N barrier and hence degrades the concentration and the mobility of the 2DEG. Although the

excessive Al can be absorbed by the barrier metal X, the thickness ratio of Ti/Al is critical to optimize the contact resistance of Ti/Al/X/Au ohmic contacts. Q. Feng, et al. [148] and X. Kong, et al. [149] reported an optimal Ti/Al thickness ratio of 1/6 on AlGaN/GaN despite the slight differences of AlGaN barrier. For more groups, the optimized Ti/Al thickness is 1/4 although the AlGaN barrier varies in these literatures [138] [139] [150]. In short, the Ti/Al thickness in between 1/6 to 1/4 is widely accepted as the optimal ratio that yields lowest contact resistance value for Ti/Al/X/Au ohmic contact.

Table 4-1 Benchmarking of the resistance values and the process conditions of conventional Ti/Al/X/Au ohmic contacts on AlGaN/GaN HEMTs

Metal Stack	Substrate	R_c (Ω.mm)	ρ_c (Ω.cm²)	RTA	Ref
Ti/Al/Pt/Au (20/80/40/150 nm)	AlGaN/GaN	0.039	5.38×10^{-8}	900 °C, 35s	[138]
Ti/Al/Ni/Au (20/120/40/50 nm)	Al _{0.26} GaN/GaN	0.12	n.a.	825 °C, 35s	[68]
Ti/Al/Ni/Au (12/200/40/100 nm)	n-Al _{0.27} GaN/GaN	0.17	7.7×10^{-8}	900°C, 30s	[140]
Ti/Al/Ni/Au (15/220/40/50 nm)	Al _{0.24} GaN/GaN	0.15	2.8×10^{-6}	700°C, 60s	[69]
Ti/Al/Ti/Au (30/80/120/55 nm)	n-Al _{0.2} GaN/GaN	n.a.	6.33×10^{-6}	700°C, 30s	[147]
Ti/Al/Ni/Au (20/120/55/40 nm)	Al _{0.3} GaN/GaN	n.a.	8.4×10^{-7}	multi-step, max. 830 °C	[148]
Ti/Al/Ni/Au (20/120/55/40 nm)	Al _{0.215} GaN/GaN	0.239	9.7×10^{-7}	870 °C, 50s	[149]
Ti/Al/Nb/Au (15/60/35/50 nm)	Al _{0.3} GaN/GaN	< 0.6	$(5 - 7) \times 10^{-6}$	830-1000 °C, 30s	[150]

The resistance values and the process conditions of conventional Ti/Al/X/Au ohmic contacts on AlGaN/GaN HEMTs are benchmarked in Table

4-1. By using the Ti/Al thickness ratio of 1/6 and annealing temperature above 825 °C, R_c value below 0.3 $\Omega\cdot\text{mm}$ with a ρ_c smaller than $5\times 10^{-6} \Omega\cdot\text{cm}^2$ are achieved [68][69][148][149]. Although other ohmic metal schemes such as V/Al/Pt/Au [151] and Mo/Al/Mo/Au [152] with special surface treatment [152][153] have been reported with good performance, the ohmic contact of Ti/Al/Ni/Au is still the dominant metal scheme used in most reported gold-based ohmic contacts in AlGaIn/GaN HEMTs.

4.4 Summary of Non-gold Ohmic contacts

In order to achieve AlGaIn/GaN HEMTs with high drain current and low specific on-resistance (R_{on}), low resistance ohmic contacts with sharp edge acuity, smooth surface morphology and high thermal stability are essential [145]. Recently, AlGaIn/GaN HEMTs on large size Si substrate (up to 200 mm diameter) have been investigated in order to develop a high-volume low-cost fabrication process using the existing 200 mm Si CMOS facilities [56][154]. However, conventional Ti/Al/X/Au ohmic scheme cannot be used in a Si CMOS line because of the contamination issue caused by gold. In addition, the gold will increase the production cost which is undesirable. Therefore, the development of non-gold ohmic contacts is necessary for the CMOS-compatible fabrication of AlGaIn/GaN HEMTs.

In the past years, research on CMOS-compatible non-gold ohmic contact has intensified. For non-gold metal scheme, Ti/Al bilayer was usually used on

doped AlGa_N barrier on sapphire substrate [144][155]. In 2011, H. -S. Lee, et al. have reported a Ti/Al/W ohmic contact with R_c and ρ_c values of $0.49 \Omega \cdot \text{mm}$ and $6.5 \times 10^{-6} \Omega \cdot \text{cm}^2$ after annealing at 870°C , respectively [59]. These results were achieved with 30 nm recess of AlGa_N/Ga_N heterostructure. As the total thickness of Al_{0.26}Ga_{0.74}N barrier and UID-Ga_N cap was 19.5 nm, the recess process even partially removed the UID-Ga_N buffer. In the same year, A. Malmros, et al. announced a Ta/Al/Ta ohmic contact with an impressive R_c of $0.06 \Omega \cdot \text{mm}$ [156]. It should be pointed out that this R_c value was achieved on a Al_{*x*}Ga_{1-*x*}N barrier with a lower Al mole fraction at $x = 0.14$. The Ta/Al/Ta scheme was also tried on Al_{0.25}Ga_{0.75}N/Ga_N HEMT, resulting in a higher R_c of $0.28 \Omega \cdot \text{mm}$. In 2012, researchers from IMEC, Belgium, reported ohmic contacts using Ti/Al/Ti/TiN [56] and Ti/Al/W [57] metallizations on Al_{0.25}Ga_{0.75}N/Ga_N/Al_{0.18}Ga_{0.82}N double heterostructure. The Ti/Al/W stacks were deposited directly onto the AlGa_N surface, a better R_c value of $0.65 \Omega \cdot \text{mm}$ was achieved after 600°C annealing as compared to R_c of $1.25 \pm 0.15 \Omega \cdot \text{mm}$ for Ti/Al/Ti/TiN scheme with partial recess of the Al_{0.25}Ga_{0.75}N barrier and lower temperature annealing at 550°C . In 2013, Z. H. Liu, et al. demonstrated Au-free AlGa_N/Ga_N HEMTs device with Ti/Al/Ni/Pt ohmic contacts. The annealing at 975°C yielded a R_c of $\sim 0.6 \Omega \cdot \text{mm}$ [58]. More recently, J. -G. Lee, et al. [61] and W. H. Tham, et al. [157] have reported the Ti/Al and Ti/Al/NiV ohmic contacts, respectively. Both groups used the low temperature annealing at 550°C for the integration of gate-first fabrication. The reported non-gold ohmic contacts are listed in Table 4-2. Although significant breakthroughs have

been made, there are still shortcomings of these ohmic contacts, e.g. the complex fabrication process with AlGaIn barrier recess, high contact resistance, low reproducibility etc. Thus, there is still room to further develop and optimize the non-gold ohmic contact on AlGaIn/GaN heterostructure for CMOS-compatible AlGaIn/GaN HEMTs processes.

Table 4- 2 The reported non-gold ohmic contacts to AlGaIn/GaN HEMTs

Metal Stack	HEMT Structure	RTA (°C)	R_c ($\Omega \cdot \text{mm}$)	ρ_c ($\Omega \cdot \text{cm}^2$)	AlGaIn Recess	Ref
Ti/Al/W (60/100/30 nm)	$\text{Al}_{0.26}\text{Ga}_{0.74}\text{N}/\text{GaN}$	870	0.49	6.5×10^{-6}	Y	[59]
Ta/Al/Ta (10/280/20 nm)	$\text{Al}_{0.14}\text{Ga}_{0.86}\text{N}/\text{GaN}$	550	0.06	-	-	[156]
	$\text{Al}_{0.25}\text{Ga}_{0.75}\text{N}/\text{GaN}$	600	0.28			
Ti/Al/Ti/TiN (20/100/20/60 nm)	$\text{Al}_{0.25}\text{Ga}_{0.75}\text{N}/\text{GaN}$ / $\text{Al}_{0.18}\text{Ga}_{0.82}\text{N}$	550	1.25 (± 0.15)	-	Y	[56]
Ti/Al/W (20/100/20 nm)	$\text{Al}_{0.25}\text{Ga}_{0.75}\text{N}/\text{GaN}$ / $\text{Al}_{0.18}\text{Ga}_{0.82}\text{N}$	800	0.65	-	-	[57]
Ti/Al/Ti/TiN (20/200/20/60 nm)	$\text{Al}_{0.21}\text{Ga}_{0.79}\text{N}/\text{GaN}$	550	1.4	-	-	[7]
Ti/Al/Ni/Pt (20/60/40/50 nm)	$\text{Al}_{0.25}\text{Ga}_{0.75}\text{N}/\text{AlN}$ / GaN	975	0.6	-	-	[58]
Ti/Al (40/200 nm)	$\text{Al}_x\text{Ga}_{1-x}\text{N}/\text{GaN}$	550	0.76	-	Y	[61]
Ti/Al/NiV (20/200/70 nm)	$\text{Al}_{0.24}\text{Ga}_{0.75}\text{N}/\text{AlN}$ / GaN	550	0.8	6.3×10^{-6}	-	[157]
Ta/Si/Ti/Al/Ni/Ta (5/5/20/120/40/30 nm)	$\text{Al}_{0.26}\text{Ga}_{0.74}\text{N}/\text{GaN}$	850	~ 0.22	7.8×10^{-7}	-	This work

4.5 Non-gold Ta/Si/Ti/Al/Ni/Ta Ohmic contacts

Recently, AlGaIn/GaN HEMTs using a non-gold ohmic contact with Ta/Si/Ti/Al/Ni/Ta (8/2/15/140/30/25 nm) metallization has been reported [112]. After 30 s rapid thermal annealing at 800 °C in N₂ ambient, a R_c value of

0.24±0.07 $\Omega \cdot \text{mm}$ with a ρ_c of $1.25 \times 10^{-6} \Omega \cdot \text{cm}^2$ were achieved. Although the AlGaIn/GaN HEMT with this CMOS-compatible ohmic contact exhibited comparable Johnson's figure of merit ($J\text{-FOM} = f_T \times BV_{gd}$) with the gold-based device, systematic study on the ohmic contacts, particularly the mechanism of low ohmic resistance, has not been carried out yet. In this section, further optimization of the ohmic contacts was implemented. After performing the design of experiments with different metal schemes, an optimal metal stack of Ta/Si/Ti/Al/Ni/Ta (5/5/20/120/40/30 nm) was realised. The Ta metal was selected because it has a relative close work function ($\sim 4.0 - 4.8$ eV, thickness dependent) as compared to the electron affinity of GaN (~ 4.1 eV [129]). The Si layer was inserted in between Ta and Ti for the following considerations: (1) Si that down diffuse into AlGaIn/GaN could cause local doping at the M-S interface area, and (2) Si could react with Ta and/or Ti to form silicides, which might be helpful to generate ohmic contact. Thus the incorporation of Ta and Si is expected to further reduce R_c . In addition, both Ta and Si are CMOS-compatible metals which are commonly used in Si fabrication line.

The AlGaIn/GaN heterostructure was grown on 4-inch high resistive Si (111) substrate using MOCVD. From bottom to top, the heterostructure consists of a 1.4 μm -thick transition layer, a 0.8 μm -thick unintentionally-doped (UID-) GaN buffer, an 18 nm-thick UID-Al_{0.26}Ga_{0.74}N barrier, and a 2 nm-thick UID-GaN cap. All the III-nitride layers, including GaN cap, AlGaIn barrier, and GaN buffer, are all un-doped. At room temperature, the 2DEG sheet density is measured as $1.1 \times 10^{13} \text{ cm}^{-2}$, with a carrier mobility of $1450 \text{ cm}^2/\text{V} \cdot \text{s}$ [112]. After

the mesa isolation by BCl_3/Cl_2 dry etching (depth: ~ 150 nm), ohmic patterns were defined by the contact-mode lithography. Before the e-beam evaporation of Ta/Si/Ti/Al/Ni/Ta (5/5/20/120/40/30 nm) metal stacks, the native oxide in the open area of AlGaIn/GaN surface was removed by a buffered HF solution etching for 60 s followed by deionized water rinsing. After the metallization, the samples were rapid thermal annealed at various temperatures (650, 700, 750, 800, 850, 875, and 900 °C, separately), in N_2 ambient for 30 s. For comparison purpose, a reference sample with conventional gold-based Ti/Al/Ni/Au (20/120/40/50 nm) contacts was also prepared and annealed at 825 °C for 30 s to achieve lowest ohmic resistance [154].

Figure 4-7 shows the schematic cross-section of the fabricated (a) non-gold and (b) gold-based ohmic contacts on AlGaIn/GaN HEMTs on Si substrate. For this non-gold metal scheme, The Al-layer in this metal stack reacts with Ti forming a low resistive Al-Ti alloy. This serves the same purpose as in the case of conventional gold-based Ti/Al/Ni/Au ohmic contact [149]. The observation of Al-Ti alloy phase in XRD spectrum verifies the effectiveness of Al-layer. The Ni-layer may have prevented the out-diffusion of Al to the surface when the sample was annealed at high temperature, thus avoiding the formation of rough surface morphology. The top-Ta layer was deposited to protect the Ni metal from oxidation. After the ohmic contacts were prepared, the samples were electrically and physically characterized by various methods, including I-V measurement to obtain R_c and ρ_c , Atomic Force Microscope (AFM) to evaluate the surface morphology, X-ray diffraction (XRD) to analyze the alloy phase

formation, and Secondary Ion Mass Spectrometry (SIMS), High Resolution Transmission Electron Microscope (HR-TEM), and Energy Dispersive X-ray (EDX, including line scan and area scan) to further identify the structure and elemental species of the contact. The results of characterization will be shown and discussed in the following sections.

Figure 4-7 Schematic diagram of (a) Ta/Si/Ti/Al/Ni/Ta and (b) Ti/Al/Ni/Au ohmic contacts on AlGaN/GaN HEMTs on Si substrate

The I-V characteristics of a 5 μm channel with the channel width of 50 μm are shown in Figure 4-8. The inset shows the optical image of measured patterns. During the measurement, the DC voltage varies in the range of −5 V to 5 V, with an incremental step of 0.2 V. The I-V curve of 650 °C annealed sample is not given here as it exhibits non-linear behavior. The linear I-V characteristics, as observed on samples annealed at 700 °C and above, suggest the ohmic behavior with annealing at higher temperature. With the change of annealing temperature from 700 °C to 850 °C, the slope of linear I-V curve gradually increases, indicating the decreasing trend of the total resistance. The

further increase of annealing temperature until 900 °C does not yield distinctive differences in the slope of linear I-V curve, which means the R_c values of ohmic contacts annealed at 850 °C, 875 °C and 900 °C are very close to each other if the R_{sh} variation is negligible. The gold-based ohmic contact still exhibits slightly smaller contact resistance, as shown in Figure 4-8, the slope of linear I-V curve of Ti/Al/Ni/Au contacts is larger than that of other I-V curves.

Figure 4-8 I-V characteristics of the 5 μm channel (width: 50 μm)

R_c and ρ_c values of non-gold and gold-based contacts are shown in Figure 4-9. When the annealing temperature is elevated from 700 °C to 850 °C, R_c drastically reduces from 18.12 $\Omega\cdot\text{mm}$ to 0.22 $\Omega\cdot\text{mm}$ with a decrease of ρ_c from $6.3\times 10^{-3} \Omega\cdot\text{cm}^2$ to $7.8\times 10^{-7} \Omega\cdot\text{cm}^2$ accordingly. These values are consistent with the trend of I-V slope change in Figure 4-9. When the annealing temperature was extended to 875 °C and 900 °C, an insignificant increase for

both R_c and ρ_c is observed.

Figure 4-9 R_c and ρ_c values of Ta/Si/Ti/Al/Ni/Ta contacts at different annealing temperatures and of conventional Ti/Al/Ni/Au contacts annealed at 825 °C

In order to test the reproducibility of the non-gold ohmic contacts, more samples are prepared at the annealing temperature range 850 - 900 °C and characterized. The corresponding R_c and ρ_c values are summarized in Table 4-3. The excellent repeatability across the samples as well as across optimal annealing temperatures is observed, as almost all R_c values fall into the range of 0.2 - 0.3 $\Omega \cdot \text{mm}$. The results reveal that the non-gold Ta/Si/Ti/Al/Ni/Ta ohmic contacts have a lowest mean R_c ($\sim 0.22 \Omega \cdot \text{mm}$, 850 °C annealing, sample #2) as compared to other published values (see Table 4-2). Compared to the 0.28 $\Omega \cdot \text{mm}$ obtained on $\text{Al}_{0.25}\text{Ga}_{0.75}\text{N}/\text{GaN}$ HEMT, our R_c/x_{Al} yields a value of 0.85, which is lower than the value 1.12 reported by A. Malmros, et al [156]. Also, a

slight improvement has been achieved (R_c reduction from $0.24 \Omega \cdot \text{mm}$ to $0.22 \Omega \cdot \text{mm}$, with ρ_c from $1.25 \times 10^{-6} \Omega \cdot \text{cm}^2$ to $0.78 \times 10^{-6} \Omega \cdot \text{cm}^2$) if compared to our own report with same metallization and different layer thicknesses [112].

Table 4-3 Statistics of R_c and ρ_c values calculated from different TLM patterns in the samples annealed at different temperatures

Annealing Temp.	Resistance	Measurements					Mean $\pm$ SD
850 °C #1	$R_c (\Omega \cdot \text{mm})$	0.23	0.25	0.24	0.26	-	0.25 $\pm$ 0.02
	$\rho_c (\times 10^{-6} \Omega \cdot \text{cm}^2)$	1.06	1.30	1.17	1.47	-	1.25 $\pm$ 0.15
850 °C #2	$R_c (\Omega \cdot \text{mm})$	0.25	0.17	0.20	0.25	-	0.22 $\pm$ 0.03
	$\rho_c (\times 10^{-6} \Omega \cdot \text{cm}^2)$	0.99	0.47	0.67	0.97	-	0.78 $\pm$ 0.22
850 °C #3	$R_c (\Omega \cdot \text{mm})$	0.28	0.29	0.28	0.24	-	0.27 $\pm$ 0.02
	$\rho_c (\times 10^{-6} \Omega \cdot \text{cm}^2)$	1.53	1.66	1.62	1.00	-	1.45 $\pm$ 0.27
875 °C	$R_c (\Omega \cdot \text{mm})$	0.24	0.25	0.25	0.25	-	0.25 $\pm$ 0.01
	$\rho_c (\times 10^{-6} \Omega \cdot \text{cm}^2)$	1.84	1.36	1.40	1.31	-	1.33 $\pm$ 0.06
900 °C	$R_c (\Omega \cdot \text{mm})$	0.25	0.22	0.29	0.28	-	0.25 $\pm$ 0.02
	$\rho_c (\times 10^{-6} \Omega \cdot \text{cm}^2)$	1.33	1.06	1.75	1.67	-	1.36 $\pm$ 0.22
Ti/Al/Ni/Au @ 825 °C	$R_c (\Omega \cdot \text{mm})$	0.13	0.19	0.18	0.25	0.15	0.18 $\pm$ 0.04
	$\rho_c (\times 10^{-6} \Omega \cdot \text{cm}^2)$	0.31	0.81	0.69	1.31	0.47	0.72 $\pm$ 0.34

For the purpose of surface morphology observation for non-gold and gold-based contacts, both optical microscope observation and AFM scanning are carried out. The AFM scanning is executed on a $5 \times 5 \mu\text{m}^2$ square shape area for all the samples. At least four different locations on each sample are measured to test the uniformity of surface topography. The scan frequency is 0.5 Hz to avoid noise effects. Figure 4-10 gives statistics of surface root mean square (RMS) value, as well as images of non-gold and gold-based contact surface under optical microscope. The RMS surface roughness of as-deposited

contacts is around ~ 3 nm. After annealing, the RMS value of annealed surface slightly increases to ~ 5.5 nm. The annealed surfaces are quite uniform, as almost all RMS data are in the range of $\sim 5 - 6$ nm. More impressively, the surface morphology is found to be independent of the annealing temperature.

Figure 4-10 Statistics of RMS surface roughness and optical images of non-gold and gold-based contact

In different publications, the surface roughness of conventional Ti/Al/Ni/Au ohmic contacts exhibits a large variation (RMS $\sim 84 - 165$ nm [158][149]). Our gold-based contact yields a surface RMS = 62.3 nm. Even compared to the relatively smoother surface of Ti/Al/Ni/Au contacts in this study, the non-gold Ta/Si/Ti/Al/Ni/Ta contacts exhibit a 12 times improvement in surface morphology. The surface morphology difference is also observable even under optical microscope, as shown in the inset of Figure 4-10. The

significant improvement in surface morphology is due to the presence of top-Ta layer. Owing to its high melting point (3017 °C), the refractory top-Ta is effective in preventing the out-diffusion of other metals, and hence maintaining the smooth surface morphology of the non-gold ohmic contact after annealing.

Figure 4-11 shows comparison of non-gold as-deposited, non-gold 850 °C-annealed, and gold-based 825 °C-annealed contacts by means of AFM 2D- and 3D-images. The peak-valley distances of these three samples are 39.8 nm, 50 nm, and 407 nm, respectively. Compared to the Ti/Al/Ni/Au ohmic contact, the non-gold contact has a significantly improved surface morphology, which is desirable for uniform current flow and small signal attenuation.

Figure 4-11 Surface 2D- and 3D-images of (left) as-deposited, (middle) 850 °C-annealed non-gold ohmic contacts, and (right) 825 °C-annealed gold-based ohmic contacts

To investigate the mechanism of the low contact resistance of the

Ta/Si/Ti/Al/Ni/Ta ohmic contacts with smooth surface morphology, the annealed and as-deposited samples are selectively characterized by powder-mode XRD scanning. The powder-mode is selected because the annealed metal consists of various alloys and the intensity signal of small-grain-size alloy could be very low. The X-ray wavelength in our system is 1.539 Å. The scanning is carried out in the 2θ range of $20 - 100^\circ$ with a very low scan speed (1° per minute with an angle incremental step of 0.01°). Figure 4-12 shows the XRD spectrum of as-deposited and annealed ohmic contacts at different temperatures. The prominent peaks of as-deposited metal stacks are GaN, Si, Al and Al-Ni alloy. The small amount Al-Ti alloy is reasonable as the 40 nm-thick Ni layer is deposited exactly onto the 120 nm-thick Al layer. The alloy phase could be a result of metal diffusion. After thermal annealing, the new XRD peaks are mainly piled up at the 2θ range of $25^\circ - 37^\circ$ and $71^\circ - 77^\circ$.

The phase analysis for XRD spectrum reveals that these compounds mainly include metal alloys (e.g. Al_3Ni_2 , Al_3Ni , Al_3Ti , etc), titanium-silicide (e.g. Ti_5Si_4 , TiSi_2 , TiSi , etc), and aluminum oxide (Al_2O_3 , mainly). The reason for the presence of Al_2O_3 is unclear, as the metal evaporation and rapid thermal annealing are executed in vacuum and N_2 ambient, respectively. L. Wang, et al. [159][160] also observed the formation of Al_2O_3 after annealing. At and above 850°C , no additional XRD peaks are found, which indicates that no additional alloy phase formation occurs. Besides the stabilized Ti_5Si_4 (203) phase at $2\theta \sim 34^\circ$, the XRD intensity of TiSi_2 (004) and TiSi (610) peaks at around 95° increases with the increase of annealing temperature. From this, we believe that

the observation of low R_c is due to the formation of Ti_xSi_y alloy.

Figure 4-12 XRD spectrum of selective annealed samples and as-deposited sample

The as-deposited and annealed contacts are selectively characterized by HR-TEM as well as EDX line-scan. Since the contact will be inevitably damaged during the Focused Ion Beam (FIB) cutting, the investigation should be arranged at the final stage of the whole characterization. The cross-sectional HR-TEM images of (a) as-deposited, (b) 750 °C-annealed, and (c) 850 °C-annealed contacts are presented in Figure 4-13. The interface areas of annealed sample are shown by zoomed-in graphs. The as-deposited sample in Figure 4-13 (a) exhibited clear metal layers with relatively smooth interfaces. The measured thicknesses values for Ta/Si/Ti/Al/Ni/Ta ($\sim 4.6/4.5/20/117/40/28$ nm) are in good agreement with the designed values (5/5/20/120/40/30 nm).

Figure 4-13 Cross sectional HR-TEM images of (a) as-deposited, (b) 750 °C-annealed, (c) 850 °C-annealed contacts, and (d) and (e) zoomed interface areas of 850 °C-annealed contact in (c), and (f) zoomed interface area of 750 °C-annealed contact in (b)

The 750 °C- and 850 °C-annealed contacts display a similar cross-sectional structure with distinctive layers (see Figure 4-13 (b) and (c)). No significant change is found for top-Ta after annealing. Due to the high melting point, the top-Ta is effective in preventing the surface morphology from further degradation. Under the top-Ta, both annealed samples had an approximate 13-nm-thick Al_xO_y layer. The presence of Al_xO_y here is consistent with the phase analysis of XRD spectrum (see Figure 4-12) as well as with the publications of I. Adesida's team [160][159]. The mechanism of Al_xO_y formation is not

confirmed yet, because the metal deposition and thermal annealing are executed in vacuum and N₂ ambient, respectively. However, the bright ribbon area in the middle of as-deposited top-Ta metal (see Figure 4-13 (a)) has high concentration of oxygen, which could be a possible oxygen source. More noteworthy, the ohmic behavior is not degraded by the insulating Al_xO_y, because large amount of Ta nanoparticles are incorporated into the oxide layer. These Ta nanoparticles could form conductive paths, thus facilitating the transport of electrons (see Figure 4-13 (d)).

The next two layers below in the annealed contact, sequentially, are Al-Ni alloy and Al-Ni-Ti alloy. Figure 4-14 gives the EDX line scan profile of (a) as-deposited, (b) 750 °C-annealed, and (c) 850 °C-annealed samples. The scan location refers to the vertical line with a cross end in each TEM image on the left side. The semi-quantitative EDX analysis approximates the two alloy layers as Al₃Ni₂ and Al₃NiTi, respectively. These two thick layers occupy the majority thickness of annealed contact. The presence of Al_xO_y and the abnormal high concentration of oxygen in top-Ta are verified by EDX line-scan profile again. For as-deposited contact, the inter-diffusion of Al with adjacent Ni and Ti is identified in Figure 4-14 (a), which conforms to the detection of Al-Ni alloy in as-deposited metal stacks by XRD in Figure 4-12. The prominent difference between the 750 °C- and 850 °C-annealed contacts is the interfacial layer sandwiched in the metal alloy and the AlGaIn/GaN heterostructure. During thermal process, the Ti reacts with Si then to form Ti-silicide. At the interface of 850 °C-annealed contact, the Ti_xSi_y has intermixed with the bottom-Ta (see

Figure 4-14 EDX line-scan profiles of (a) as-deposited, (b) 750 °C-annealed, and (c) 850 °C-annealed contacts

Figure 4-13 (e)). On the contrary, the crystallized Ti_xSi_y is clearly separated from amorphous bottom-Ta in 750 °C-annealed contact (see Figure 4-13 (f)).

Most of the Ti-silicides have relative low metal work function in the range of 3.69 eV to 4.18 eV [161], which almost match or are even smaller than the electron affinity of GaN (~ 4.1 eV [129]). A low energy barrier is formed in the M-S interface when Ti_xSi_y penetrates through bottom-Ta, and directly contacts with AlGaIn/GaN, resulting in an enhanced electron transport by the thermionic emission mechanism. That is the case of 850 °C-annealed contact. However, when the annealing temperature is not high enough to support the downward diffusion of Ti_xSi_y , the energy barrier at the M-S interface is still provided by Ta although Ti_xSi_y starts sintering since 500 °C [161]. Again, the intermixing and separating of Ti_xSi_y with bottom-Ta in 750 °C- and 850 °C-annealed contacts are revealed in Figure 4-14 (b) and (c), respectively. From EDX line-scan results, the low energy barrier due to diffused Ti_xSi_y is verified. Another possible mechanism of resistance decreasing with elevated annealing temperature is the TiSi_2 phase transfer from high-resistive C49 to low-resistive C54 at 700 °C [162]. Although TiSi_2 is not the only Ti-silicide at the interface, the phase change would be helpful to achieve low contact resistance.

Ta-silicide is not observed although Ta is in direct contact with Si in as-deposited stacks. The phenomenon can be explained as Ta-silicide formation needs higher temperature (above ~ 600 °C) [163]. When 600 °C is achieved in the annealing chamber, the limited amount of Si has been depleted by excessive Ti. Similar to the conventional gold-based contact, the Al-layer is applied to

consume Ti, then to form low electrical resistive [149] and high oxidation resistive [155] Al-Ti alloy. The Ni-layer in this study may have prevented the out-diffusion of Al to the surface when the sample is annealed at high temperature, thus preventing the formation of rough surface topography. Obviously, besides maintaining smooth surface, the top-Ta is also effective in protecting the other metals from being oxidized.

4.6 Conduction Mechanism of Ta/Si/Ti/Al/Ni/Ta Ohmic contact

To facilitate the design and the optimization of ohmic contact, it is useful to have a good understanding of the ohmic metal formation as well as the conduction mechanism. For gold-based ohmic contacts, such studies have been reported. The formation of the ohmic contact with low R_c is mainly due to the inter-diffusion of gold and N-atoms, which forms a low resistivity Al-Ti-N alloyed layer and results in a Thermionic-Field Emission (TFE) dominated electron transport at the M-S interface [164][165]. In contrast, although non-gold ohmic contacts have been published with improved R_c in recent years [56][57][59][156], their ohmic metal formation processes and the conduct mechanisms were not well understood. Based on the non-gold Ta/Si/Ti/Al/Ni/Ta ohmic contact we developed with a superior low $R_c \sim 0.22 \Omega \cdot \text{mm}$, the conduction mechanism is systematically investigated using the temperature dependent I-V measurement.

The samples fabricated previously have been damaged by the FIB cutting for HR-TEM characterization. Hence, new samples were prepared to study the conduction mechanism as well as to verify the reproducibility of developed Ta/Si/Ti/Al/Ni/Ta non-gold ohmic contacts in different batches. The fabrication procedures have been described in Section 4.5. Two annealing temperatures (850 °C and 900 °C) were selected for rapid thermal annealing because the lowest R_c values were observed in the annealing temperature range $850\text{ °C} \leq T \leq 900\text{ °C}$, as given by Figure 4-8, Figure 4-9 and Table 4-3. A reference sample with Ti/Al/Ni/Au (20/120/40/50 nm) ohmic contacts was prepared on a similar epi-wafer for comparison purpose. For all the samples, R_c , R_{sh} and ρ_c values were extrapolated using TLM patterns at the temperature range of -25 °C to 150 °C , with a step increment of 25 °C .

Figure 4-15 shows R_{total} versus the TLM gap length at different measurement temperatures for the 850 °C-annealed sample. The measured R_{total} data points can be perfectly linear fitted at all measurement temperatures. For the sample annealed at 900 °C, similar trend was observed (The figure is not shown). Figure 4-16 gives ρ_c values (calculated using *Eqn 4.7*) of the ohmic contact versus measurement temperature. R_c and ρ_c of all samples at 25 °C are given in Table 4-4. Again, the non-gold ohmic contacts exhibited low R_c in all measured TLM patterns at annealing temperature of 850 °C and 900 °C, suggesting the good reproducibility of Ta/Si/Ti/Al/Ni/Ta ohmic metal scheme.

Typically, three possible models are used to explain the conduction mechanisms at an M-S interface in ohmic contacts. These are Thermionic

Emission (TE), Thermionic-Field Emission (TFE), and Field Emission (FE, or tunneling), which exhibit different temperature dependencies. Theoretically, the dominant mechanism can be identified by using the conduction model to fit the ρ_c values measured at various temperatures.

Figure 4-15 R_{total} versus gap length of a TLM pattern on 850 °C-annealed non-gold sample

Figure 4-16 gives the temperature dependent ρ_c values of gold-based and non-gold contacts with the TFE and TE fitted curves. The error bars of non-gold contacts are based on the statistical data of 3 TLM patterns at different locations for each sample. At -25 °C, the non-gold ohmic contact exhibit larger ρ_c values (850 °C: $\sim 2.3 \times 10^{-6}$, 900 °C: $\sim 3.3 \times 10^{-6}$, gold-based: $\sim 1.4 \times 10^{-6} \Omega \cdot \text{cm}^2$). With the increase of temperature, ρ_c of both non-gold samples decreases more rapidly than that of gold-based sample. At and above 100 °C, both non-gold contacts

exhibit smaller ρ_c than gold-based contacts.

Figure 4-16 Calculated ρ_c values of gold-based and non-gold (with error bar) ohmic contacts versus measurement temperature with the TE and TFE mechanism fitted curves.

The calculated mean ρ_c of non-gold ohmic contact can be well fitted using the TE model expressed by Eqn 4.11 [117]

$$\rho_c = \frac{k_B T}{q A^* T^2} \exp\left(\frac{q \Phi_B}{k_B T}\right) \quad \text{Eqn 4.11}$$

where $q\Phi_B$ denotes the Schottky barrier height at the M-S interface, A^* represents the Richardson constant, and q , k_B , T are the electron charge, the Boltzmann constant and the temperature in unit K, respectively. The Richardson constant $A^* = 4\pi m_n^* q k_B^2 / h^3$ (27.12 A/cm²·K²), in which m_n^* is the effective electron mass ($0.22m_0 = 2.06 \times 10^{-31}$ Kg) and h is the Planck constant.

The extracted $q\Phi_B$ values for the samples annealed at 850 °C and 900 °C range from $\sim 0.109 - 0.116$ eV (mean value: 0.113 eV) and $\sim 0.117 - 0.124$ eV (mean value: 0.121 eV), respectively, as listed in Table 4-4. These parameters not only verified the good uniformity of the non-gold ohmic contact across devices but were also consistent with the reported R_c values (See Table 4-3). The extrapolated mean $q\Phi_B$ values are only ~ 4.5 times of the thermal energy at room temperature ($k_B T = 0.026$ eV at $T = 300$ K). The extremely low barrier energies make the electron transport by surmounting the barrier with sufficient energy possible.

Table 4 - 4 Fitted conduction parameters ($q\Phi_B$, V_n , N_D) and measured R_c and ρ_c values at 25 °C of gold-based and non-gold ohmic contacts on un-doped AlGaIn/GaN HEMTs grown on silicon

Sample	TLM	$R_c@25^\circ\text{C}$ ($\Omega\cdot\text{mm}$)	$\rho_c@25^\circ\text{C}$ ($10^{-6}\Omega\cdot\text{cm}^2$)	$q\Phi_B$ (eV)	Cond. Model
850 °C	A1	0.22	0.82	0.109	<u>TFE</u>
	A2	0.25	1.09	0.113	
	A3	0.25	1.06	0.116	
	Mean	0.24	0.99	0.113	
900 °C	B1	0.27	1.34	0.124	
	B2	0.28	1.37	0.117	
	B3	0.29	1.45	0.123	
	Mean	0.28	1.39	0.121	
Gold-based	C1	0.18	0.86	0.233	TFE*

*For the gold-based contact, fitted $V_n = 0.008$ eV and $N_D = 8.42 \times 10^{18} \text{ cm}^{-3}$ in TFE model

Using a same measurement at the temperature range of -113 °C to 27 °C,

D. -W. Kim, et al. reported the TE mechanism of 700 °C annealed Ti/Si/Ti/Au ohmic contacts on n-GaN ($7 \times 10^{16} \text{ cm}^{-3}$). The low energy barrier at the M-S interface was also attributed to the formation of low work function Ti_xSi_y [161]. The reason for the slight increase in R_c (from ~ 0.24 to $\sim 0.28 \text{ } \Omega \cdot \text{mm}$) and in $q\Phi_B$ at 900 °C could be the instability of TiSi_2 at high annealing temperature. N. de Iannarolle, et al. have realized the decomposition of formed alloy TiSi_2 into TiN and Si when annealed at $\sim 925 \text{ } ^\circ\text{C}$ in N_2 ambient [166]. This possibly explains the slight degradation of our non-gold ohmic contacts annealed at 900 °C.

Several conduction mechanisms for gold-based ohmic contacts in GaN HEMTs have been postulated recently [145][149][167][168]. These include (1) Tunneling conduction due to heavily n-doped region resulting from the out-diffusion of Ga- and N-atoms [145][167]; (2) Direct link due to the penetration of conductive TiN alloy through AlGaN barrier until 2DEG [149]; (3) Low Schottky barrier resulting from the formation of metal alloys [168]. However, no clear conclusion has been made except for the widely acknowledged tunneling component due to the large concentration of N-vacancies. Z. H. Liu, et al. proposed a TFE based mechanism for the annealed Ti/Al/Ni/Au ohmic contact annealed at 850 °C [165]. In addition to the TFE mechanism for Ti/Al/Ni/Au ohmic contacts annealed at 750°C, F. Iucolano, et al. even observed a ‘metal-like’ ($\rho_c \propto T^{1.8}$ relationship) behavior for the sample subjected to 850 °C annealing [164]. In our study, the calculated ρ_c values of gold-based contacts exhibited a distinctively different trend compared to the

non-gold contacts as shown in Figure 4-16, suggesting a different type of conduction mechanism. The ρ_c versus temperature plots can be fitted well using the TFE conduction model given in Eqn 4.12 [169], in which V_n denotes the energy difference of Fermi level and conduction band, the calculation of E_{00} has been shown in Eqn 4.1, and E_0 is expressed using Eqn 4.13

$$\rho_c = \frac{1}{qA^*} \cdot \frac{k_B^2}{\sqrt{\pi(\Phi_B + V_n)E_{00}}} \cdot \cosh\left(\frac{E_{00}}{k_B T}\right) \times \sqrt{\coth\left(\frac{E_{00}}{k_B T}\right) \cdot \exp\left(\frac{q\Phi_B + V_n}{E_0} - \frac{V_n}{k_B T}\right)} \quad \text{Eqn 4.12}$$

$$E_0 = E_{00} \coth\left(\frac{E_{00}}{k_B T}\right) \quad \text{Eqn 4.13}$$

In Eqn 4.1, the dielectric constant $\varepsilon = 8.9\varepsilon_0$. The TFE model fitted N_D , $q\Phi_B$ and V_n values of gold-based contact are given in Table 4-4. Compared to the parameters in Ref [165], a much lower $q\Phi_B$ (0.233 eV) and a similar N_D ($8.42 \times 10^{18} \text{ cm}^{-3}$) are obtained. This is consistent with the observed smaller R_c value for our gold-based sample. Unlike TE which has a strong dependence on temperature, TFE typically exhibits a weaker temperature dependency because it is a combination of TE and FE mechanism. For FE which is mainly due to tunneling, it is almost independent of temperature.

The schematic diagram of the band structure can be used to explain the formation processes and the conduction mechanism of ohmic contacts on AlGaIn/GaN HEMTs. Figure 4-17 shows the band diagram of as-deposited metal stack in contact with GaN cap/AlGaIn/GaN heterostructure. As the work functions of first metal layers are close to each other ($q\Phi_m(\text{Ti}) \sim 4.33 \text{ eV}$, $q\Phi_m$

(Ta) ~ 4.26 eV [117]), the band structure of gold-based and non-gold samples can be qualitatively illustrated by Figure 4-17 (a). Before annealing, both contacts are non-ohmic due to the thick Schottky barrier at the M-S interface.

Figure 4-17 The band diagram of (a) as-deposited metal stacks (b) annealed gold-based ohmic contacts with TFE mechanism, and (c) annealed non-gold ohmic contacts with TE mechanism onto the AlGaN/GaN

Figure 4-17 (b) shows the band diagram of TFE model for 850 °C

annealed Ti/Al/Ni/Au contact proposed by Z. H. Liu, et al [165]. The three regions of the contact included two barriers and a high conductive layer. The first barrier (region I) is located at the metal and the modified GaN/AlGaN interface, and the second barrier (region III) is formed between the modified GaN/AlGaN and the 2DEG. The term “modified GaN/AlGaN” (region II) was used because the material properties of GaN/AlGaN sandwiched in between the two barriers had been significantly modified. This is the prominent change in the band diagram of gold-based Ti/Al/Ni/Au ohmic contacts after thermal annealing. The chemical reaction (mainly Ti with GaN cap/AlGaN) caused large amount of N-atoms diffusion, and hence resulted in the region II with high density of N-vacancies. These N-vacancies can act as n-type dopant with which the tunneling conduction was greatly enhanced.

In contrast, our non-gold ohmic contacts yielded a totally different band diagram, as shown in Figure 4-17 (c). The annealing helped with the formation of metal alloys. At the same time, no reaction between the metal and the GaN cap layer was observed. This is confirmed by the HR-TEM and area EDX mapping in Figure 4-18. As the AlGaN/GaN heterostructure was well separated by the bottom refractory Ta layer, the Ti did not react with III-nitrides thus the GaN cap/AlGaN/GaN heterostructure was kept intact. The most notable change during annealing was at the M-S interface. When the Ti_xSi_y was alloyed and diffused downwards and contacted the GaN cap/AlGaN interface, the Fermi level was re-aligned due to the smaller work function of Ti_xSi_y [161]. Subsequently, a small barrier height value (0.113 eV for 850 °C annealed

contacts) was from the TE model fitting (See Table 4 - 4). A band bending of GaN cap at the interface was assumed. Because of the dipole effect, small amount of electrons could be gathered at the GaN cap surface [170]. However, the localized electrons and band bending would not affect the electron transport because the thickness value (2 nm) of GaN cap is very small. Compared to the Figure 4-17 (b), the high conductive region (II) due to high density N-vacancies which appears in gold-based ohmic contact does not exist in the non-gold ohmic contacts, by which an alternative mechanism can be inferred.

The HR-TEM and area EDX characterizations were used to explain the ohmic contact formation between metals and the GaN cap/AlGaIn/GaN epilayers. As shown in the inset of Figure 4-18 (a), the narrow dark layer at the M-S interface of non-gold contacts annealed at 850 °C consists of bottom Ta and Ti_xSi_y alloy. Because of a 5 nm-thick bottom Ta layer, the rest of the metal (Si/Ti/Al/Ni/Ta) cannot diffuse into the GaN-cap/AlGaIn/GaN heterostructure. The boundary of M-S interface is clear in the whole scanned area, as shown in Figure 4-18 (a). The EDX elementary mapping in Figure 4-18 (b) indicates that there is no out-diffusion of N-atoms from the GaN-cap/AlGaIn/GaN into the ohmic metals as no traces of elementary N_2 species are detected in the metal alloy. Without the heavily doped region to assist FE conduction, we can thus attribute the conduction mechanism to be mainly TE.

In most gold-based ohmic contacts, the penetration of metal alloy (e.g. metallic TiN) into GaN/AlGaIn was observed [171]. As a result, the AlGaIn barrier was damaged even partially consumed [171][172]. The interfacial

reaction caused large amount of out-diffusion of N-atoms [145][149][167], which has been observed in EDX area mapping [145] or in Auger Electron Spectroscopy (AES) line scanning [148][161]. The tunneling conduction was enhanced by the N-vacancies left behind [145][172]. This explains the FE component in the conduction of gold-based ohmic contacts as the TFE can be regarded as a combination of TE and FE [165][164].

Figure 4-18 (a) The cross-sectional view of HR-TEM image, and (b) the EDX elementary mapping of 850 °C-annealed non-gold ohmic contacts [The EDX scan area is marked in (a) with dashed border].

4.7 Summary

A CMOS-compatible non-gold ohmic contact has been developed using the Ta/Si/Ti/Al/Ni/Ta scheme with thickness combination of 5/5/20/120/40/30 nm. This non-gold ohmic contacts exhibit smooth surface morphology (RMS $\sim$ 5.5 nm), relatively large temperature window of rapid thermal annealing ($\sim$ 850 - 900 °C), good reproducibility, and most importantly, low contact resistance (min. R_c and ρ_c are $\sim$ 0.22 $\Omega \cdot \text{mm}$ and $\sim 7.8 \times 10^{-7} \Omega \cdot \text{cm}^2$, respectively, for 850 °C annealing). The characterization results verify that the low R_c is contributed by the formation of low work function Ti_xSi_y as well as the mixing of Ti_xSi_y with bottom Ta during the rapid thermal annealing. The extracted ρ_c values of Ta/Si/Ti/Al/Ni/Ta ohmic contacts show a stronger temperature dependency as compared to those of the conventional Ti/Al/Ni/Au ohmic contacts, which can be well fitted using the thermionic emission (TE) model with small energy barrier heights of 0.113 eV and 0.121 eV for the annealing temperature at 850 °C and 900 °C, respectively. On the contrary, the current transport of gold-based ohmic contact is dominated by thermionic-field emission (TFE) with an energy barrier of 0.233 eV at the M-S interface and an n-type doping of $8.42 \times 10^{18} \text{ cm}^{-3}$ in AlGaIn barrier caused by interfacial reaction. Since the downward diffusion of metal into GaN cap/AlGaIn/GaN heterostructure or the out-diffusion of N-atoms from III-nitride into metal alloy are not observed, the TE conduction in the non-gold ohmic contact is well supported by HR-TEM and EDX.

Chapter 5 Sputtered TiN Schottky Gate

Schottky gate refers to a Metal-Semiconductor (M-S) contact with a non-negligible Schottky barrier height (SBH or $q\Phi_B$) at the interface. A rectifying current-voltage behavior is typically observed for a Schottky gate as the electron transport from metal to semiconductor is more difficult than the transport in a reverse direction. The Ni/Au bi-layer Schottky gate for conventional AlGaIn/GaN fabrication cannot be used in Si CMOS line as gold will bring contamination. Hence, the development of CMOS-compatible Schottky gate is necessary to realize the fabrication of AlGaIn/GaN HEMTs in existing Si foundries. In this chapter, the Ni/Au bi-layer gate will be reviewed in the first section. Next, the development of non-gold sputtered TiN gate will be presented. The characterization of SBH by various methods, the analysis on the improved SBH, and the HEMT device characteristics with sputtered TiN gate will be discussed in detail. Finally, the gate leakage mechanism is studied systematically.

5.1 Conventional Ni/Au bi-layer gate

Theoretically, the SBH of a Schottky contact on AlGaIn/GaN HEMT can be determined by the Schottky-Mott theory [128]. Metals with high work function, such as Au (5.1 eV), Pd (5.12 eV), Ni (5.15 eV), Pt (5.65 eV), etc., are commonly used for the Schottky contact in both GaN and AlGaIn [173]

[174][175] to maximize the SBH which in return can reduce the reverse leakage current and the OFF-state power consumption of AlGaIn/GaN HEMT.

Table 5-1 Summary of SBH ($q\Phi_B$) and ideality factor (n) of large work function metals on AlGaIn/GaN HEMT structure

Metal	Substrate	$q\Phi_B$ (eV)	n	Ref
Au	Al _{0.27} Ga _{0.73} N/GaN	1.05 (I-V-T)	~ 1.36	[176]
	Al _{0.3} Ga _{0.7} N/GaN	~ 1.6 (I-V)	-	[177]
Pd	Al _{0.27} Ga _{0.73} N/GaN	1.08 (I-V-T)	~ 1.35	[176]
Ni-based	Al _{0.25} Ga _{0.75} N/GaN	0.99@320K (I-V)	1.81	[178]
	Al _{0.26} Ga _{0.74} N/GaN	0.74 (I-V)	-	[174]
	Al _{0.3} Ga _{0.7} N/GaN	~ 1.42 (I-V)	~ 1.25	[177]
	Al _{0.27} Ga _{0.73} N/GaN	0.98 (I-V-T)	~ 1.36	[176]
	Al _{0.25} Ga _{0.75} N/GaN	1.27 (C-V)	-	[179]
	Al _{0.23} Ga _{0.77} N/GaN	0.51 (I-V)	-	[180]
	Al _{0.29} Ga _{0.71} N/GaN	0.82 (I-V)	~ 1.77	[181]
	Al _{0.15} Ga _{0.85} N/GaN	1.31 (PE [‡])	-	[182]
Pt-based	Al _{0.27} Ga _{0.73} N/GaN	0.98 (I-V-T)	~ 1.37	[176]
	Al _{0.42} Ga _{0.58} N/GaN	0.84 (SRPS [§])	2.69	[183]
	n-Al _{0.3} Ga _{0.7} N/GaN	1.61@298K (I-V)	-	[184]
	Al _{0.25} Ga _{0.75} N/GaN	0.8-1.02@RT (I-V)	-	[185]
	Al _{0.2} Ga _{0.8} N/GaN	1.05 (I-V)	1.1 - 1.3	[186]
	Al _{0.26} Ga _{0.74} N/GaN	0.72 (I-V)	-	[174]

[‡] PE - internal photoemission

[§] SRPS - synchrotron radiation photoemission spectroscopy

The reported values of SBH ($q\Phi_B$) and ideality factor (n) of large work function metals on AlGaIn/GaN HEMT structure are summarized in Table 5-1. For most of the cases in Table 5-1, the Al mole fraction is in between 0.2 to 0.3 except for the 0.42 and 0.15 in Ref [183] and Ref [182], respectively, which is

also the most commonly used Al ratio range for HEMT device fabrication. However, even for the similar $\text{Al}_x\text{Ga}_{1-x}\text{N}/\text{GaN}$ HEMT structure (e.g. 0.25 - 0.27) with same contact metal (e.g. Ni-based contact [174][178][176][179] and Pt-based contact [174][176][185]), the measured SBH exhibits large variations (Ni-based: 0.74 - 1.31 eV, Pt-based: 0.72 - 1.02 eV), which could be possibly caused by the difference of the characterization methods, by the AlGa_N/Ga_N surface conditions, or by the pre-treatment or post-treatment of Schottky contact deposition, etc.

For the device fabrication of AlGa_N/Ga_N HEMT, the electron beam (e-beam) evaporated Ni/Au bi-layer gate [6][68][85][181][187][188][189][190] is most commonly used because of the large metal work function of Ni and the high oxidation resistivity of Au. The thickness of Ni and Au are typically ~ 20 - 50 nm and ~ 100 - 400 nm, respectively, to accommodate the subsequent metal lift-off process.

To improve the Schottky characteristics of Ni-based contacts on AlGa_N/Ga_N HEMTs, some treatments on the semiconductor or on the gate metals were studied. N. H. Kim, et al. reported that the active states at the M-S interface can be passivated by post-deposition annealing at 400 °C [181]. R. Chu, et al. observed a three orders reduction on the magnitude of reverse leakage by CF₄ surface plasma treatment prior to the Ni/Au gate deposition [189]. By optimizing the treatment using BCl₃ plasma followed by CF₄ plasma, the SBH was even slightly increased (0.6 to 0.7 eV) [188]. The leakage reduction in both articles was attributed to the formation of F-containing

surface layer, by which the vertical leakage path is blocked by an insulating layer [188][189]. N. -H. Lee, et al. demonstrated a three-order of leakage reduction by SF_6 and in-situ N_2 plasma treatment. More significantly, the subthreshold swing (SS) is reduced from ~ 150 to ~ 70 mV/dec, suggesting a sharper ON/OFF switching [190]. D. Qiao, et al. tried optimizing the SBH of Ni by varying the Al mole fraction using bulk $\text{Al}_x\text{Ga}_{1-x}\text{N}$. The continuous increase in SBH as well as the decrease of n are observed with x from 0 to 0.17 [74].

5.2 Sputtered TiN Schottky Gate

In order to have a gate process which is Si CMOS-compatible, the conventional Ni/Au bi-layer gate must be replaced by other non-gold materials because: (1) Au is not allowed in Si foundries due to contamination issue, (2) the usage of Au will inevitably increase the production cost, and (3) Ni is difficult to be selectively removed by plasma etching. Hence, alternate gate materials are necessary to realize a CMOS-compatible fabrication process. Compared to other reported Ni/Au-free Schottky gates such as Ti [177], Cr [177], Ir [179], Re [179], ITO [180], TaN [180], W[191], WN_x [191], RuO_x [192], graphene [193], etc., the sputtered TiN [62][194][195][196][197] exhibits advantages such as: (i) The metal work function ($q\Phi_m$) of sputtered TiN is ~ 5 eV [198], which is ~ 0.9 eV higher than χ_s (GaN) ~ 4.1 eV [129]; (ii) TiN is compatible for Si CMOS process. It can be deposited by sputtering [62][195]

[196][197], which is the preferred deposition technique for Si CMOS foundries. It can also be removed using dry etching by fluorine-based plasma or using wet etching by diluted HF; (iii) TiN itself has many attractive merits such as high thermal stability [196][198], low electrical resistivity ($\sim 55 \mu\Omega\cdot\text{cm}$, smaller than pure Ti film) [199], low cost, etc.

The implementation of TiN as the Schottky gate on AlGaIn/GaN HEMTs structure has drawn attention in recent years. T. Kawanago, et al. observed lower current collapse, lower gate leakage and lower frequency dispersion by using a TiN gate [62]. J. -P. Ao, et al. realized that the TiN on AlGaIn/GaN HEMT is able to withstand up to 600 °C for 1 hour or up to 850 °C for several minutes [196]. A. Soltani, et al. observed an increase of SBH (0.4 eV to 0.95 eV) in AlGaIn/GaN HEMTs with TiN after thermal annealing at 600 °C [195]. Y. -H. Lin, et al. demonstrated comparable DC characteristics of HEMT using TiN/Cu gate as compared to the device with conventional Ni/Au gate [197]. However, the above-mentioned works mainly focused on the improvement of device characteristics such as current collapse [62], thermal stability [195][196] [197][200], high voltage stress effect [197], etc. As one of the most important parameters to assess the performance of a Schottky contact, the SBH of sputtered TiN on AlGaIn/GaN HEMTs has not been fully characterized thus far. In addition, the gate leakage current mechanism and energy levels of the interface states of sputtered TiN on AlGaIn/GaN HEMTs also have not been studied to date. The leakage current mechanism was commonly observed to associate with the semiconductor material. For example, Schottky emission was

realized in n-type Ge substrates with Pt [201] and Se [202] Schottky contacts and Poole-Frenkel emission was widely adopted for the III-nitrides (GaN [203], AlGaN/GaN [20][94][204], and AlInN/GaN [204][205]) with Ni-based Schottky contacts by electron beam (e-beam) evaporation. However, TiN gate was commonly deposited by sputtering process [62][195][196][197][200][206][207], which may lead to plasma damages to the surface of AlGaN/GaN HEMT structure. Hence, it is important to study and understand the gate leakage current mechanism of AlGaN/GaN HEMTs with sputtered TiN. Motivated by the promising material properties and the insufficient study of TiN, the characterization of Schottky parameters and the analysis of gate leakage current mechanism will be focused in this chapter.

5.3 Experimental details

The AlGaN/GaN HEMT structure was grown on a 4-inch high-resistivity Si (111) substrate by Metal Organic Chemical Vapor Deposition (MOCVD). At room temperature, the measured 2DEG carrier density, electron mobility and sheet resistance are $\sim 9.8 \times 10^{12} \text{ cm}^{-2}$, $\sim 1450 \text{ cm}^2/\text{V} \cdot \text{s}$ and $440 \text{ } \Omega/\square$. The device fabrication, including Schottky barrier diodes (SBDs) and HEMTs, started from mesa isolation which is formed by BCl_3/Cl_2 plasma etching to a depth of $\sim 150 \text{ nm}$. Then, ohmic metal stacks are e-beam evaporated with Ti/Al/Ni/Au (20/120/40/50 nm) followed by rapid thermal annealing at $775 \text{ }^\circ\text{C}$ for 30 s in N_2 ambient. The contact resistance (R_c) of ohmic contacts was measured at ~ 0.3

$\Omega\cdot\text{mm}$. Next, 200 nm-thick TiN is sputtered at a DC power of 450 W and an $\text{N}_2/(\text{N}_2+\text{Ar})$ gas flow rate of 40/70 (unit: sccm). The pre-pressure and the deposition pressure are below 5×10^{-6} Torr and maintained at $\sim 7.7\times 10^{-3}$ Torr, respectively. The other flow rates [$\text{N}_2/(\text{N}_2+\text{Ar}) = 10/70$ and $70/70$ (No Ar gas)] were also used to study the influence of gas flow to the Schottky parameters. After a lift-off process, the devices were then passivated with a 120 nm-thick Si_3N_4 layer by Plasma Enhanced Chemical Vapor Deposition (PECVD) at 300 °C.

Figure 5-1 shows (a), (c) the schematic cross-sectional diagrams and (b), (d) the optical microscope images of the guard-ring type SBDs and AlGaIn/GaN HEMTs with sputtered TiN gate, respectively. The fabricated TiN SBDs were characterized by temperature dependent current-voltage (I-V-T) and capacitance-voltage (C-V) measurements to study the Schottky parameters and the gate leakage current mechanism. The characteristics of single-gate AlGaIn/GaN HEMT with sputtered TiN gate were also demonstrated and compared with the conventional Ni/Au-gate device. For HEMTs, both the source-gate distance (L_{gs}) and the gate length (L_{g}) are 2 μm . The gate width (W_{g}) is 80 μm . The gate-drain distance (L_{gd}) had different values (2, 4, 6, 8, 10 μm) which would be specified with the discussion of measured data/figures. All the above-mentioned measurements were performed using an Agilent B1505A power device analyzer.

Figure 5-1 (a) (c) Cross-sectional schematic diagram and (b) (d) microscope image of Schottky barrier diode and single-gate AlGaIn/GaN HEMT with sputtered TiN gate.

5.4 Characterization of Schottky Barrier Height (SBH)

The I-V characteristics of an ideal SBD is given by [208],

$$\begin{aligned}
 I &= I_s \left[\exp\left(\frac{qV_D}{nk_B T}\right) - 1 \right] = I_s \left\{ \exp\left[\frac{q(V - IR_s)}{nk_B T}\right] - 1 \right\} \\
 &\approx I_s \exp\left(\frac{qV_D}{nk_B T}\right) \quad (qV_D \gg k_B T) \quad \text{Eqn 5.1}
 \end{aligned}$$

in which the saturation current I_s can be expressed by

$$I_s = AA^*T^2 \exp\left(-\frac{q\Phi_B}{k_B T}\right) \quad \text{Eqn 5.2}$$

In Eqn 5.1 and Eqn 5.2, n is the ideality factor, V_D is the voltage drop across the diode, R_s is the series resistance, A and A^* denote the anode area and the Richardson constant, respectively, q is the unit electron charge, k_B is the Boltzmann constant, and T is the temperature in unit K.

The extraction of SBH using the forward I-V characteristics is a commonly used method with straightforward calculation. Using the I-V method, the influence of R_s in Eqn 5.1 is neglected. Once I_s value is obtained from the intercept of linear fit of logarithmic I - V to the I - axis, the $q\Phi_B$ can be simply derived using Eqn 5.2 with the known A and A^* .

Figure 5-2 (a) Forward I-V characteristics, and (b) Extracted $q\Phi_B$ and n value by I-V method of TiN SBDs.

The typical I-V characteristics of a fabricated SBD are shown in Figure 5-2 (a). The measurement is carried out from the temperature of 290 K until 340 K, with a step incremental of 10 K. The increase of the current under forward bias with increasing temperature reveals that the thermionic emission is still the dominant mechanism in forward bias [202]. Figure 5-2 (b) gives the $q\Phi_B$ and n values at each temperature, which are calculated using *Eqn 5.2*. The error bars are based on the data of six SBDs with the same anode diameter. The increase of $q\Phi_B$ and decrease of n with the increasing measurement temperature are observed, suggesting the presence of other conduction mechanism rather than the thermionic emission and/or the inhomogeneous nature of the M-S interface [209]. At room temperature (300 K), the the mean value of $q\Phi_B$ is 1.126 ± 0.012 eV with the n of 1.76 ± 0.07 . Comparatively, the Ni/Au Schottky gate yields a $q\Phi_B \sim 0.96$ eV ($n \sim 1.45$) using the same I-V measurement (The figure is not shown).

For comparison purpose, two different I-V-T measurement methods are used to determine $q\Phi_B$ of sputtered TiN on AlGaIn/GaN HEMT structure. The first method is the activation energy plot. By using the activation energy plot, the *Eqn 5.2* should be rewritten as

$$\ln\left(\frac{I_s}{AT^2}\right) = \ln A^* - \frac{q\Phi_B}{k_B T} \quad \text{Eqn 5.3}$$

Apparently, the slope and the intercept of linear fit of $\ln(I_s/AT^2)$ vs. $1/k_B T$ gives the $-q\Phi_B$ and $\ln A^*$, respectively, as shown in Figure 5-3. The error bars are summarized from the values of six TiN SBDs. The extracted $q\Phi_B$ is 1.098 eV,

which is highly consistent to the value obtained by I-V method. However, the calculated A^* from the intercept ($2.61 \text{ A/cm}^2\text{K}^2$) is far from its theoretical value of $26.4 \text{ A/cm}^2\text{K}^2$ [176]. Abnormal A^* values were also observed by other researchers. P. Hacke, et al. reported an even smaller A^* of $0.006 \text{ A/cm}^2\text{K}^2$ for gold Schottky contact on n-type GaN using the same activation energy plot [210]. J. D. Guo, et al. obtained A^* of 6.61 and $0.44 \text{ A/cm}^2\text{K}^2$ for Pt and Pd Schottky contacts on n-type GaN, respectively [211]. The large discrepancies in fitted A^* and theoretical A^* are primarily due to the fitting error, and the fitted value may not be used in other experiments [136].

Figure 5-3 Linear fit of $\ln(I_s/AT^2)$ versus $1/k_B T$ and the extracted $q\Phi_B$ and A^* value of TiN SBDs using the activation energy method.

Another method related to I-V-T measurement is the modified Norde's function [129]. Due to the presence of R_s , the voltage drop across the SBD (V_D)

is not equal to the applied voltage (V) upon the whole measurement circuit in Eqn 5.1. Obviously, the I_s value obtained from the linear fit of logarithmic $I - V$ is not reliable as the influence of R_s is overlooked [212]. Another more accurate method involving $I-V-T$ measurement is the modified Norde's function, by which the effect of R_s can be completely eliminated. The modified Norde's function is given by [129][213]

$$F(V) = \frac{qV}{2k_B T} - \ln\left(\frac{I}{T^2}\right) \quad \text{Eqn 5.4}$$

By defining a function $G(T)$ where

$$G(T) = 2F(V) + (2 - n) \ln\left(\frac{I}{T^2}\right) \quad \text{Eqn 5.5}$$

when $F(V)$ is at its minimum $F_{\min}(V)$, $G(T)$ can be rewritten as,

$$G(T) = 2 - n[\ln(AA^*) + 1] + \frac{n}{k_B T} q\Phi_B \quad \text{Eqn 5.6}$$

When A and n are known parameters, the value of $q\Phi_B$ and A^* can be derived from the slope and the intercept of the linear fit of $G(T)$ vs. $n/k_B T$, respectively. In Figure 5-2 (b), n value continuously decreases with the increase of measurement temperature. In order to confine the n variation and use the average n in the calculation, the measurement was carried out at a relatively narrow temperature range (290 to 340 K, as shown in Figure 5-2 (a)).

Figure 5-4 (a) shows the representative $F(V)$ vs. V plot of a TiN SBD. The linear fit of $G(T)$ vs. $1/k_B T$ is given in Figure 5-4 (b). The error bars in Figure 5-4 (b) are calculated from statistical value of six TiN SBDs. The extracted effective $q\Phi_B$ is 1.105 eV, which is well agreed by the 1.126 ± 0.012 eV by $I-V$ and 1.098 eV by activation energy plot, suggesting that R_s does not

play a significant role in the forward conduction of TiN SBDs. However, similar to the observation in the activation energy plot, the fitted $A^* \sim 6.36$ $\text{A}/\text{cm}^2\text{K}^2$ by modified Norde's function is much smaller than the theoretical value of at $26.4 \text{ A}/\text{cm}^2\text{K}^2$ [176].

Figure 5-4 (a) $F(V)$ vs. V plot, and (b) linear fit of $G(T)$ vs. $1/k_B T$ and extracted $q\Phi_B$ and A^* value of TiN SBDs using modified Norde's function

The large discrepancy, which is also observed by other groups using modified Norde's function [129][136][213], can be explained by the influence of process parameters, such as metal type, thicknesses, deposition method, etc [214]. A. C. Schmitz, et al. attributed the A^* variation to the changed interface by the diffusion and/or intermixing of metal and GaN [136]. In addition, we believe that the statistical error may play a significant role in getting a much smaller A^* values [136]. Since the measurement temperature range is far from

the $G(T)$ - axis ($T \rightarrow +\infty$), any minimal $G(T)$ variation (which may be caused by measurement error) will result in a large fluctuation on the intercept of linear fit to $G(T)$ - axis. Nevertheless, the slope will not be influenced considerably. In other words, the modified Norde's function is more appropriate to extract $q\Phi_B$ rather than A^* . The extracted effective $q\Phi_B$ and A^* of Ni/Au from the modified Norde's function is ~ 0.86 eV and ~ 0.022 A/cm²K², respectively (The figure is not shown).

The C-V measurement can also be used to calculate $q\Phi_B$ value. However, the conventional $1/C^2$ vs. V plot is not appropriate for our study because: (1) the AlGaIn/GaN HEMT structure is not a bulk semiconductor with uniform doping; (2) the depletion width and built-in voltage cannot be simply calculated using the abrupt-junction approximation [208]. A different physical model should be applied for the C-V behavior of SBD on the HEMT structure [179]. Based on the quantitative analysis of piezoelectric and spontaneous polarization, the sheet carrier concentration $n_s(x)$ of 2DEG is expressed as [179][81]

$$n_s(x) = \frac{\sigma(x)}{q} - \frac{\epsilon_0 \epsilon_s(x)}{q^2 d} [q\Phi_B - E_F(x) - \Delta E_C(x)] \quad Eqn 5.7$$

In Eqn 5.7, x denotes the Al mole fraction in $Al_xGa_{1-x}N$; When $x = 0.26$, the bound sheet charge $\sigma(x)/q$ and the dielectric constant $\epsilon_s(x)$ of $Al_xGa_{1-x}N$ are estimated to be $1 \times 10^{13}/\text{cm}^2$ and 10.322, respectively [81]. The polarization effect of GaN cap was ignored as its thickness is only 2 nm; ϵ_0 is the vacuum permittivity; d denotes the $Al_xGa_{1-x}N$ barrier thickness (18 nm, see Figure 5-2); the band offset $\Delta E_C(x)$ [81][215] and the Fermi level $E_F(x)$ [179] are given by:

$$\Delta E_C(x) = 0.7[E_g(x) - E_g(0)] \quad Eqn 5.8$$

$$E_F(x) = E_0(x) + \frac{\pi \hbar^2}{m^*(x)} \cdot n_s(x) \quad Eqn 5.9$$

where $E_g(x)$ is the band gap of $Al_xGa_{1-x}N$ [81][215],

$$E_g(x) = 6.13x + 3.42(1 - x) - x(1 - x) \quad Eqn 5.10$$

$\hbar$ is reduced Planck constant; the effective electron mass $m^*(x) = 0.22 m_0$ [179];

the 2DEG ground sub-band level $E_0(x)$ is calculated using [179]

$$E_0(x) = \left[\frac{9\pi \hbar q^2 n_s(x)}{8\varepsilon_0 \varepsilon_s(x) \sqrt{8m^*(x)}} \right]^{\frac{2}{3}} \quad Eqn 5.11$$

Once $n_s(x)$ is known, $q\Phi_B$ can be extracted since the other terms of Eqn 5.7 are mathematically calculable.

Figure 5-5 gives the C-V characteristics of TiN SBD at a frequency of (a) 1 MHz and (b) 100 kHz by sweeping voltage successively forward (anode bias from -6 to 0 V) and reverse (anode bias from 0 to -6 V). The hysteresis effect, expressed by a shift from reverse sweep to forward sweep, is almost not observable, suggesting the electron charging and discharging by trap centers at the M-S interface are not significant [178]. The dispersion of frequency is relatively small (saturated capacitance: ~ 0.3 and $\sim 0.315 \mu F/cm^2$ at 1 MHz and 100 kHz, respectively), suggesting the low leakage current of TiN Schottky gate [62]. The zero bias $n_s(x)$ can be obtained from the integration of C-V curve [215] [216], given by

$$n_s = \int_{V_{th}}^0 \frac{C}{qA} dV \quad Eqn 5.12$$

Figure 5-5 The forward and reverse C-V characteristics of TiN SBD at the measurement frequency of (a) 1 MHz and (b) 100 kHz.

The threshold voltage (V_{th}) can be determined by the intercepts of two linear fits of channel charge Q with the bias [216]. The extraction of V_{th} is shown in Figure 5-6. As indicated in the inset of Figure 5-6, n_s value is calculated as $5.775 \times 10^{12} \text{ cm}^{-2}$. Accordingly, $E_F(x)$ and $\Delta E_C(x)$ in Eqn 5.7 are obtained as 0.245 eV and 0.359 eV, respectively, resulting in a $q\Phi_B$ of 1.446 eV. The C-V measurement is carried out on six SBDs to examine the uniformity. With the integrated n_s ranging from 5.643 to $5.859 \times 10^{12} \text{ cm}^{-2}$, $q\Phi_B$ is calculated in between 1.411 to 1.486 eV, suggesting that TiN contacts on AlGaIn/GaN HEMT structure are quite uniform across SBDs. Comparing to $q\Phi_B \sim 1.1$ eV obtained by I-V and I-V-T measurements, the C-V method (1.435 ± 0.029 eV) yields a ~ 0.33 eV larger $q\Phi_B$ value. The variation in $q\Phi_B$ is commonly

observed by using different characterization techniques [129][136][217]. The presence of oxide layer at the M-S interface may contribute to the discrepancy [217]. Moreover, the estimation error of $\sigma(x)/q$ may introduce inaccuracies into the subsequent calculation, influencing the magnitude of $q\Phi_B$ in return.

Figure 5-6 The charge Q in the 2-dimensional channel of AlGaIn/GaN HEMT structure (inset) The sheet carrier concentration (n_s) integrated from C-V curve

The reported $q\Phi_B$ values for Ni/Au-free and other TiN-based Schottky contacts on $\text{Al}_x\text{Ga}_{1-x}\text{N}/\text{GaN}$ HEMTs are summarized in Table 5-2. Except for a few materials such as WN_x [191] and Re [179], the sputtered TiN exhibits larger $q\Phi_B$ than the rest of the Ni/Au-free Schottky contacts. Compared to other reported Schottky contacts using TiN-based stacks, our sputtered TiN exhibits higher $q\Phi_B$ value [194][195][197]. This improvement could be explained by three reasons.

Table 5-2 Summarized values of SBH ($q\Phi_B$) of Ni/Au-free and TiN-based Schottky contacts onto $\text{Al}_x\text{Ga}_{1-x}\text{N}/\text{GaN}$ HEMT structure ($0.2 \leq x \leq 0.32$)

Materials	Substrate	$q\Phi_B$ (eV)	Ref
Ti	$\text{Al}_{0.2}\text{Ga}_{0.8}\text{N}/\text{GaN}$	~ 0.95 (I-V)	[177]
Cr		~ 0.92 (I-V)	
ITO	$\text{Al}_{0.23}\text{Ga}_{0.27}\text{N}/\text{GaN}$	0.62 (I-V)	[180]
TaN		0.54 (I-V)	
WN_x	$\text{Al}_{0.25}\text{Ga}_{0.75}\text{N}/\text{GaN}$	max. 1.21 (I-V)	[191]
W		0.89 (I-V)	
Ir	$\text{Al}_{0.25}\text{Ga}_{0.75}\text{N}/\text{GaN}$	1.12 (C-V)	[179]
Re		1.68 (C-V)	
graphene	$\text{Al}_{0.25}\text{Ga}_{0.75}\text{N}/\text{GaN}$	0.4 (I-V)	[193]
RuO_x	$\text{Al}_{0.24}\text{Ga}_{0.76}\text{N}/\text{GaN}$	~ 0.82 (I-V)	[192]
TiN/Cu	AlGaN/GaN	~ 0.67 (I-V)	[197]
TiN/Ti/Al/Ti/TiN	$\text{Al}_{0.25}\text{Ga}_{0.75}\text{N}/\text{GaN}$	~ 0.68 (I-V)	[194]
TiN/Au	$\text{Al}_{0.32}\text{Ga}_{0.68}\text{N}/\text{GaN}$	max. 0.95 (I-V)	[195]
Ni/Au	$\text{Al}_{0.26}\text{Ga}_{0.74}\text{N}/\text{GaN}$	~ 0.96 (I-V)	This Work
		~ 0.86 (I-V-T)	
TiN		~ 1.1 (I-V(-T)) ~ 1.44 (C-V)	

Firstly, C. L. Hinkle, et al. has reported a ~ 0.55 eV increase on the effective work function of TiN with the incorporation of oxygen ($[\text{O}] = 2.8 \times 10^{21} \text{ cm}^{-3}$ near the interface) [218]. The incorporated oxygen was also detected in our sputtered TiN using Rutherford backscattering spectrometry (RBS), as shown in Figure 5-7. The RBS measurements were carried out using 1 mm collimated 2 MeV He^+ . Backscattered particles were detected using a surface barrier detector placed at an backscattering angle of 165° [219]. Surface peaks associated with Ti and N appear at channel numbers 614 and 280, respectively. Rutherford universal manipulation program (RUMP) was used to

extract the composition and thickness of the TiN [219]. It was found that the sputtered film contained a small fraction of O which appears at around channel number 322. It is to be noted that the Ti, N and O peaks overlap with the backscattering signals from Al and Ga atoms present in the substrate. The composition of sputtered TiN was found to be 44.5 %, 44.5 % and 11 % for Ti, N and O, respectively. The oxygen compositional ratio of 11% corresponds to $[O] = 4.96 \times 10^{21} \text{ cm}^{-3}$, which is slightly larger than the concentration of oxygen in Ref [218].

Figure 5-7 RBS spectrum of sputtered TiN film on AlGaIn/GaN HEMT structure

The presence of oxygen in the sputtered TiN was also verified by the line-scan energy dispersive X-ray spectroscopy (EDX), as shown in Figure 5-8. The scanning was performed along the solid arrow in the inserted Transmission Electron Microscopy (TEM) image. Approximately 15% oxygen was uniformly distributed in the whole layer of sputtered TiN, which is consistent to the RBS

result in Figure 5-8. The oxygen impurities were due to the residual gases of the sputter chamber as the pre-vacuum was not sufficiently high. The observation of oxygen in sputtered TiN was also reported by other material researchers [220].

Figure 5-8 Line-scan EDX profile and (inset.) cross-sectional TEM image of sputtered TiN onto AlGaIn/GaN HEMT structure with SiN_x passivation

Secondly, the work function of TiN is reported to be influenced by the N₂/(N₂+Ar) ratio during the sputtering process, which in return affects the $q\Phi_B$ at the M-S interface. In order to investigate the effect of gas flow rate on the work function of sputtered TiN, three more samples were fabricated using the same processes. The N₂/(N₂+Ar) gas rates were set at 10/70, 40/70 and 70/70 (no Ar gas) (unit: sccm), respectively. Figure 5-9 shows the statistical $q\Phi_B$ and n values measured from six SBDs on each sample by I-V method. The n value is not significantly changed with the variation of N₂/(N₂+Ar). At N₂/(N₂+Ar) =

10/70 and 40/70 (unit: sccm), the extracted mean $q\Phi_B$ are ~ 1.16 and ~ 1.13 eV, respectively, suggesting that (i) the Schottky parameters of sputtered TiN on AlGaIn/GaN HEMTs at $N_2/(N_2+Ar) = 40/70$ (unit: sccm) is repeatable as compared with the extracted $q\Phi_B$ in Figure 5-2, and (ii) the variation of N_2 ratio in the total gas flow from 10/70 to 40/70 (unit: sccm) has negligible influence to the TiN work function. When $N_2/(N_2+Ar)$ was increased to 70/70 (unit: sccm), $q\Phi_B$ reduced to ~ 0.88 eV.

Figure 5-9 Extracted SBH ($q\Phi_B$) and n value of TiN on AlGaIn/GaN HEMT structure at different $N_2/(N_2+Ar)$ gas flow rates by I-V method

J. Westlinder, et al. observed a sudden enhancement of TiN work function (~ 0.7 eV) when $N_2/(N_2+Ar)$ was changed from 0.107 to 0.145. The further increase of $N_2/(N_2+Ar)$ until 0.167 did not influence the TiN work function significantly [221]. On the contrary, Y. Liu, et al. reported that the TiN work

function was also decreased (from 4.82 to 4.71 eV) with the increase of $N_2/(N_2+Ar)$ ratio from 0.17 to 0.83 [222], which is in good agreement with our results (See Figure 5-9). The deposited TiN with the flow rate of $N_2/(N_2+Ar) = 40/70$ (unit: sccm) can be correlated with the work function of TiN ($\sim 4.8 - 4.9$ eV) [221][222]. This could be a possible reason for the improved $q\Phi_B$ on our TiN/AlGaIn/GaN Schottky barrier diodes.

At last, the crystal orientation is also a factor that determines the work function of sputtered TiN. H. F. Dadgour, et al. suggested that the metal surface orientation with densely packed atoms, such as (111), was more preferred in forming the Schottky contacts with large $q\Phi_B$ than the surface orientations with loosely packed atoms, such as (100) and (110) [223]. J. Hu, et al. observed a larger $q\Phi_B$ of physical vapor deposited (PVD) TiN with ionized metal plasma source (IMP) as compared to the standard PVD TiN. The grazing incidence X-ray diffraction (GIXRD) spectrum revealed that the IMP TiN has only (111) grain orientation but PVD TiN had both (200) and (111) orientations [224]. The X-ray diffraction (XRD) spectrums of sputtered TiN at different N_2 ratios are given in Figure 5-10. The incident angle XRD scanning was set low in order to collect the X-ray reflections only from TiN film and also to prevent the X-ray reflections from the substrate. The sputtered TiN with the flow rate of $N_2/(N_2+Ar) = 10/70$ (unit: sccm) exhibited (111), (200) and (220) reflections, which were also observed by D. Dictus, et al. in PVD and IMP TiN [225]. When the N_2 ratio increased to 40/70 (unit: sccm) then to 70/70 (unit: sccm), the peak of (111) reflection disappeared and the peak of (200) reflection

increased gradually with insignificant change on the peak of (220) reflection, which is in good agreement with the observation of large $q\Phi_B$ from the TiN deposited with 40/70 (unit: sccm) flow rate [223][224].

Figure 5-10 XRD spectrum of sputtered TiN films with different $N_2/(N_2+Ar)$ gas flow ratios

5.5 AlGaIn/GaN HEMTs characteristics

Figure 5-11 shows the DC characteristics of AlGaIn/GaN HEMTs with sputtered TiN and conventional Ni/Au gate. The gate-drain distance (L_{gd}) is 4 μm with a gate width (W_g) of 80 μm . Table 5-3 shows the device DC parameters of AlGaIn/GaN HEMT with sputtered TiN gate and Ni/Au gate.

Figure 5-11 DC (a) output and (b) transfer characteristics of HEMTs with sputtered TiN and conventional Ni/Au gate

Table 5-3 Key parameters of DC and Pulsed I_{ds} - V_{ds} measurements on AlGaIn/GaN HEMTs with sputter TiN and Ni/Au gate

Parameters	TiN	Ni/Au
$I_{ds, \max}$ (mA/mm)	464	506
V_{th} (V)	-3.4	-3.6
Subthreshold Swing (mV/dec)	106.4	102.4
$g_{m, \max}$ (mS/mm)	149	146
static- R_{on} ($m\Omega \cdot cm^2$)	0.429	0.391
dynamic- R_{on} ($m\Omega \cdot cm^2$)	0.487	0.414
dynamic- R_{on} /static- R_{on}	1.14	1.06

Compared to the device with Ni/Au gate, the maximum source-drain current ($I_{ds, \max}$) of TiN gate HEMT is decreased by $\sim 8.3\%$ at the gate bias (V_g) of +1 V, as shown in Figure 5-11 (a). The maximum transconductance ($g_{m, \max}$), the threshold voltage (V_{th}) and the subthreshold swing (SS) of TiN gate HEMT are

quite close to those of Ni/Au gate device. The gate-source leakage of TiN device at pinch-off region is reduced by approximately one order of magnitude, as shown in Figure 5-11 (b).

Figure 5-12 (a) and (b) show the OFF-state breakdown voltage (BV_{gd}) characteristics of AlGaIn/GaN HEMTs measured at $V_g = -5$ V. BV_{gd} of

Figure 5-12 OFF-state breakdown voltage characteristics of AlGaIn/GaN HEMTs with (a) Ni/Au gate and (b) sputtered TiN gate; (c) BV_{gd} at different L_{gd} , and (d) two terminal gate-leakage current characteristics of AlGaIn/GaN HEMTs

AlGaIn/GaN HEMTs at the current compliance of 1 mA/mm was found to increase with the increase of L_{gd} , suggesting the breakdown was mainly caused by the lateral leakage for both TiN gate and Ni/Au gate devices. The sputtered TiN gate HEMT exhibited improved BV_{gd} of $\sim 48 - 58$ V when it was compared to the Ni/Au gate HEMT with the same L_{gd} , as shown in Figure 5-12 (c). The improvement can be explained by the lower reverse gate leakage current (I_{gd}). As illustrated in Figure 5-12 (d), I_{gd} of TiN device was approximately reduced by an order of magnitude, which is primarily contributed to the enhanced $q\Phi_B$ of sputtered TiN Schottky contact when compared to the Ni/Au Schottky contact on AlGaIn/GaN HEMT structure. The observation of the lower two-terminal gate leakage current is in good agreement with three-terminal gate leakage current (See Figure 5-11 (b))

Figure 5-13 shows the pulsed I_{ds} - V_{ds} characteristics of AlGaIn/GaN HEMTs with Ni/Au and TiN gate. The gate-lag quiescent bias of $(-5, 0)$ V was applied to both types (Ni/Au and TiN gate) of AlGaIn/GaN HEMTs. The measurement pulse period is 500 ms with a pulse width of 500 μ s. The static and dynamic on-resistance (R_{on}) values were determined from the DC and pulsed I_{ds} - V_{ds} characteristics, respectively. The obtained parameters are listed in Table 4-3. For the case of AlGaIn/GaN HEMT with TiN gate, the ratio of dynamic- R_{on} /static- R_{on} is slightly increased ($\sim 8\%$) from 1.06 to 1.14 when it is compared with the conventional Ni/Au gate.

Figure 5-13 Pulsed I-V characteristics of HEMTs with (a) conventional Ni/Au gate and (b) sputtered TiN gate.

5.6 Gate Leakage Current mechanism analysis

5.6.1 Gate Leakage current mechanism study using SBD

Figure 5-14 shows the reverse current density (J_R) - reverse bias (V_R) characteristics of sputtered TiN SBDs as a function of measurement temperatures. In Reg. I, J_R increases exponentially with the increase of V_R and becomes saturated after V_R is increased beyond -3.2 V. The saturated J_R region is defined as Reg. II. The reverse current saturation behavior was associated with a complete depletion of 2DEG channel. J_R increases with the increase of measurement temperature in both Reg. I and Reg. II. Similar observations have also been realized by other reports [203][204].

Figure 5-14 Reverse current-voltage ($J_R - V_R$) characteristics of SBD with sputtered TiN anode at various temperatures

In a Schottky contact, the reverse leakage current before saturation (Reg. I in Figure 5-14) can be explained using either Schottky (SC) emission or Poole-Frenkel (PF) emission [226]. For Schottky emission, the carriers absorb thermal energy and then emit over the potential barrier at the M-S interface, whereas in Poole-Frenkel emission, the carrier transport occurs through trap states with the assistance of the applied field [202]. J_R contributed by these two emission models are given by [227]

For Schottky emission

$$J_R = A^* T^2 E \exp\left[-\frac{q}{k_B T} \left(\phi_t - \sqrt{\frac{qE}{4\pi\epsilon_{s(h)}\epsilon_0}}\right)\right]$$

$$\propto T^2 \exp\left(\frac{q}{2k_B T} \sqrt{\frac{qE}{\pi\epsilon_{s(h)}\epsilon_0}}\right) \quad Eqn 5.13$$

For Poole-Frenkel emission

$$J_R = qn_0\mu E \exp\left[-\frac{q}{k_B T}(\phi_t - \sqrt{\frac{qE}{\pi\epsilon_{s(h)}\epsilon_0}})\right] \propto E \exp\left(\frac{q}{k_B T} \sqrt{\frac{qE}{\pi\epsilon_{s(h)}\epsilon_0}}\right) \quad \text{Eqn 5.14}$$

In Eqn 5.13, A^* is the Richardson constant. In Eqn 5.14, n_0 and μ denote the carrier density and the effective carrier mobility, respectively. The other parameters k_B , T , ϵ_0 , and q are the Boltzmann constant, the temperature, the permittivity in vacuum and the elementary charge, respectively. The high-frequency (optical) dielectric constant ($\epsilon_{s(h)} = 5.2$) of $\text{Al}_{0.26}\text{Ga}_{0.74}\text{N}$ is used as the trap emission which is much faster than the dielectric relaxation [228]. $q\phi_t$ is the effective barrier height for electron emission from a trapped state and E is the electric field at the M-S interface. Hence, the leakage current mechanism can be identified by analyzing the slope of $\ln(J_R/T^2)$ vs. $E^{0.5}$ for Eqn 5.13 and $\ln(J_R/E)$ vs. $E^{0.5}$ for Eqn 5.14 [201][202]. Theoretically, the slope of Poole-Frenkel emission (S_{PF}) is exactly two times of the slope of Schottky emission (S_{SC}), as expressed by

$$S_{PF} = \frac{d(J_R/E)}{d(E^{0.5})} = 2S_{SC} = 2 \frac{d(J_R/T^2)}{d(E^{0.5})} \quad \text{Eqn 5.15}$$

The electric field (E) across the barrier can be calculated using the equation $E = q(\sigma_b - n_s)/\epsilon_s\epsilon_0$ [204]. The bound charge (σ_b) at the hetero-interface, which is the sum of the piezoelectric polarization charge in the barrier and the difference between spontaneous polarization charge in the barrier and the buffer, is estimated to be $1 \times 10^{13} \text{ cm}^{-2}$ [81]. $\epsilon_s = 10.32$ is the static dielectric constant of $\text{Al}_{0.26}\text{Ga}_{0.74}\text{N}$. The 2DEG concentration (n_s) at the hetero-interface

can be extracted using the C-V characteristics of the fabricated SBD (see Figure 5-5). The calculated E as a function of reverse bias is shown in Figure 5-15, which will be used for the subsequent analysis of carrier transport mechanism.

Figure 5-15 Calculated E-field as a function of reverse bias of TiN SBD

The J_R/T^2 vs. $E^{0.5}$ characteristics at various temperatures are shown in Figure 5-16 (a). The measured S_{SC} , which was determined from the linearly fitted slope of J_R/T^2 with $E^{0.5}$, increased gradually with the increase of temperature as predicted by *Eqn 5.13*. Figure 5-16 (b) shows the extracted S_{SC} from the reverse J_R - V_R measurement as well as from the theoretical calculation. The error bars were calculated based on the statistical values of 10 measured SBDs. At all temperatures, the measured mean S_{SC} value was ~ 2.6 times of the theoretical value. Therefore, the carrier transport in Reg. I was unlikely dominated by Schottky emission.

Figure 5-16 (a) J_R/T^2 vs. $E^{0.5}$ and the linear fit to extract S_{SC} (b) S_{SC} measured from the reverse J_R - V_R characteristics and calculated from the theoretical equation

To verify the extracted value of S_{PF} , the J_R/E vs. $E^{0.5}$ characteristics is also plotted in Figure 5-17 (a). Similarly, the measured J_R/E can also be well fitted with $E^{0.5}$ and the high S_{PF} was observed at high measurement temperature. Although the measured statistical S_{PF} in Figure 5-17 (b) did not fit well with the theoretical values, the discrepancy is relatively small ($S_{PF}/\text{Theo. } S_{PF} = \sim 1.11 - 1.17$). Therefore, the conduction mechanism across the M-S interface is most likely due to Poole-Frenkel emission rather than Schottky emission.

Figure 5-17 (a) J_R/E vs. $E^{0.5}$ and the linear fit to extract S_{PF} (b) S_{PF} measured from the reverse J_R - V_R characteristics and calculated from the theoretical equation.

Using the Poole-Frenkel emission model (Reg. I), the parameter $\epsilon_{s(h)}$ is obtained by plotting S_{PF} versus $1000/T$ in Figure 5-18 (a). The extracted $\epsilon_{s(h)}$ in this work and other reported $\epsilon_{s(h)}$ values are also included in Table 5-4. To reduce the error, the mean values of S_{PF} are calculated from 10 measured SBDs. $\epsilon_{s(h)} = 5.28$ are obtained from the fitted graph (see Figure 5-18 (a)), which is consistent with the reported values of GaN (5.35) and AlN (4.77) [203]. This further supports the presence of Poole-Frenkel emission in the Reg. I of Figure 5-14 as a gate leakage current conduction mechanism.

Figure 5-18 (a) Slope S_{PF} versus $1000/T$ and (b) intercept I_{PF} versus $1000/T$ plots of SBDs with sputtered TiN anode. The linear fit is based on Poole-Frenkel emission model.

The parameter $q\phi_t$ of Poole-Frenkel emission is also calculated from the intercept (I_{PF}) of J_R/E -axis in Figure 5-17 (a) versus $1000/T$, as shown in Figure 5-18 (b). For comparison, the reported $q\phi_t$ values by other researchers are also tabulated in Table 5-4. Although this study employed the HEMT structure with similar Al mole fraction ($x = 0.25 - 0.28$) like other reports [94][203][204][229][230], our obtained $q\phi_t$ of 0.53 eV is larger than those of $\sim 0.16 - 0.34$ eV (see Table 5-4). The observation of high $q\phi_t$ may be due to the plasma damage to the surface of AlGaIn/GaN HEMT structure during the TiN sputtering process.

Table 5-4 The reported $\epsilon_{s(h)}$ and $q\phi_t$ of Poole-Frenkel emission for Schottky contacts on $Al_xGa_{1-x}N/GaN$ HEMT structure.

[$\epsilon_{s(h)}$ value: GaN(5.35), AlN(4.77) [203]]

Schottky contacts	HEMT Structure	$\epsilon_{s(h)}$	$q\phi_t$ (eV)	Remark	Ref
Ni	$Al_{0.25}Ga_{0.75}N/GaN$	~ 5.1	~ 0.30	§	[203]
Ni/Au	$Al_{0.28}Ga_{0.72}N/AlN/GaN$	-	0.3	§	[94]
Ni/Au	$Al_{0.26}Ga_{0.74}N/AlN/GaN$	-	0.17	§	[204]
Ni/Au	$Al_{0.28}Ga_{0.72}N/AlN/GaN$	-	0.35	§	[229]
Ni/Al/Ta	$Al_{0.26}Ga_{0.74}N/GaN$	-	0.33 ± 0.05	§	[230]
Ni/Au	$Al_{0.15}Ga_{0.85}N/AlN/GaN$	4.85	0.65	§1	[228]
Ni/Au	$Al_{0.15}Ga_{0.85}N/GaN$	-	0.517	§2	[231]
Ni/Au	$Al_{0.65}Ga_{0.35}N/AlN/GaN$	-	0.38	§	[232]
TiN	$Al_{0.26}Ga_{0.74}N/GaN$	5.28	0.53	‡	This work

§: Schottky contacts deposited by electron beam (e-beam) evaporation

‡: Schottky contacts deposited by sputtering

1: Surface CF_4 plasma treatment prior to the Schottky contact deposition

2: Mesa sidewall leakage. Mesa was formed by Cl_2 plasma and ICP-RIE

Other researchers have also observed high $q\phi_t$ [228][231]. C. Xu, et al. have realized that the Poole-Frenkel emission dominated leakage current across the sloped mesa side walls with $q\phi_t$ of 0.517 eV. The sloped mesa was formed by inductively coupled plasma etching by Cl_2 plasma [231]. Subsequently, an even higher $q\phi_t$ of 0.65 eV was reported by W. J. Ha, et al. with CF_4 plasma treatment prior to the e-beam evaporation of Schottky metal [228]. The observation of high $q\phi_t$ value in both works may not be related to the lower Al

mole fraction ($x = 0.15$) in AlGa_N barrier [228][231] as small $q\phi_t$ (0.38 eV) was also obtained from Ni/Au Schottky contacts on high Al mole fraction ($x = 0.65$) AlGa_N barrier [232]. The reported high $q\phi_t$ values could be related to the plasma damage rather than the Al mole fraction in the AlGa_N barrier. Recently, we have also reported an activation energy of ~ 0.513 eV which is related to the heavy ($^{131}\text{Xe}^+$) ion damages to the AlGa_N/Ga_N HEMT structure [113]. From these observations, the obtained high $q\phi_t$ of 0.53 eV is most probably coming from the plasma damage during the TiN sputtering process.

To illustrate the physical origin of $q\phi_t$ in Poole-Frenkel emission model for the Reg. I in Figure 5-14, the schematic band diagram is drawn for the AlGa_N/Ga_N HEMT structure with sputtered TiN as shown in Figure 5-19. Based on our reported SBH of ~ 1.1 eV [206], the obtained $q\phi_t$ of 0.53 eV may not be the dislocation-related traps as the commonly observed trap states at ~ 0.3 eV [233] ($\text{SBH} - 0.3 \text{ eV} \neq q\phi_t$) below the conduction band edge of AlGa_N/Ga_N does not match with our $q\phi_t$. The conduction along the threading dislocation line is also not possible because the threading dislocation related states are normally near or below the Fermi level [234]. The modified Poole-Frenkel emission as suggested by H. Zhang, et al. [203] is the most likely mechanism for our gate leakage current behavior in Reg. I. For this modified Poole-Frenkel emission, carrier transport from the TiN into the conductive dislocation related continuum of states must be via a trap state at the M-S interface rather than by direct thermionic emission (See Figure 5-19). The trap state could neither be too low nor too high. If the trap energy level is

The diagram illustrates the energy levels and carrier transport in a TiN/AlGaN/GaN heterostructure. The TiN layer is on the left, followed by the AlGaN layer, and the GaN layer is on the right. The conduction band edge is labeled E_c and the Fermi level is labeled E_f . The energy difference between the conduction band edge and the Fermi level in the AlGaN layer is labeled $q\Phi_t$ and is approximately 0.53 eV. The energy difference between the conduction band edge and the Fermi level in the GaN layer is labeled SBH and is approximately 1.1 eV. The diagram shows the formation of a Schottky barrier (SBH) at the AlGaN/GaN interface. The energy levels in the TiN layer are shown at a higher potential than the AlGaN layer. The energy levels in the GaN layer are shown at a lower potential than the AlGaN layer. The diagram also shows the formation of a Schottky barrier (SBH) at the AlGaN/GaN interface. The energy levels in the GaN layer are shown at a lower potential than the AlGaN layer. The diagram also shows the formation of a Schottky barrier (SBH) at the AlGaN/GaN interface. The energy levels in the GaN layer are shown at a lower potential than the AlGaN layer.

In the Reg. II of Figure 5-14, J_R is saturated with V_R but increases with the increase of measurement temperature, suggesting a thermally activated process with an $\exp(-E_A/k_B T)$ functional dependence. Figure 5-20 shows the Arrhenius plot of J_R to extract the activation energy (E_A). At $V_R = -5$ and -15 V, the extracted E_A is ~ 0.115 eV with negligible differences. Similar E_A values have also been reported by other researchers. Y. Chen et al. and E. J. Miller et al. reported E_A of ~ 0.129 eV and ~ 0.18 eV for $\text{Al}_{0.65}\text{Ga}_{0.35}\text{N}/\text{GaN}$ HEMT structure [232] and for GaN [234], respectively. A. Fontserre, et al. also observed an E_A value for $\text{Al}_{0.28}\text{Ga}_{0.72}\text{N}/\text{GaN}$ at ~ 0.4 eV in the measurement temperature range of 450 - 600 K. At lower temperature, the exhibited E_A value is smaller

than ~ 0.4 eV in this work [94].

Figure 5-20 Arrhenius plots of SBD with sputtered TiN anode at the reverse bias (a) $V_R = -5$ V and (b) $V_R = -15$ V. The activation energies (E_A) are extracted by linear fit.

Figure 5-21 illustrates the schematic band diagram in Reg. II of AlGaIn/GaN HEMT structure with sputtered TiN in Reg. II. When the reverse bias is large enough to saturate the leakage current (Reg. II: -20 V < V_R < -3.2 V), the electrons are thermally emitted to E_A then tunnel through the AlGaIn barrier [234]. At high reverse bias, the electron emission from TiN to the dislocation-related continuum through the trap states of 0.53 eV is not possible. The direct tunneling is the dominant mechanism as the effective thickness of barrier is reduced due to the high electric field across the AlGaIn layer.

Figure 5-21 Schematic band diagram of the AlGaN/GaN HEMT structure with sputtered TiN at large reverse bias (Reg. II: $-20 \text{ V} < V_R < -3.2 \text{ V}$).

5.6.2 OFF-state gate leakage current and breakdown voltage of AlGaN/GaN HEMTs

The temperature dependent gate-leakage current (I_g) and source-drain current (I_{ds}) are measured from three-terminal AlGaN/GaN HEMT at OFF-state to correlate the device characteristics with reverse leakage current mechanisms. Figure 5-22 shows high-voltage drain-biased I_g - V_{ds} characteristics of AlGaN/GaN HEMT at different temperatures. To avoid hard damage to the device, the gate voltage $V_g = -7 \text{ V}$ and maximum source-drain bias $V_{ds} = 40 \text{ V}$ are imposed for the measurements. A positive temperature dependency of I_g is observed, which is associated to the temperature assisted tunneling conduction [235].

Figure 5-22 I_g - V_{ds} characteristics of AlGaIn/GaN HEMT with sputtered TiN gate for the gate voltage $V_g = -7$ V at different temperatures (280, 300, 320, 350, 380, and 420 K)

The measured I_{ds} and I_g values at $V_{ds} = 40$ V for different temperatures are given in Figure 5-23 (a). The gradually increasing trend of I_{ds} with the increase of temperature is almost the same as that of I_g except for the difference in the magnitude of current. Figure 5-23 (b) shows the Arrhenius plot to extract the activation energy of both I_g and I_{ds} measured at OFF-state. The activation energy of 0.159 eV and 0.175 eV are fitted for I_g and I_{ds} , respectively. Similar activation energies have also been reported by other researchers. W. S. Tan, et al. extracted an $E_A \sim 0.21$ eV from I_g in the temperature ranging from 293 to 473 K (20 to 200 °C) [236]. S. Arulkumaran, et al. observed three different E_A values by the I_g and I_d measurements for the temperature range of

293 to 673 K (20 to 400 °C) [235], in which $E_A \sim 0.2$ eV was attributed to the mechanism of surface hopping conduction, same as reported in Ref [236]. The hopping conduction is associated to the surface states created by dangling bonds and/or defects or contaminations in between gate and drain.

Figure 5-23 (a) Source-drain (I_{ds}) and gate-leakage (I_g) current as a function of measurement temperature and (b) Activation energy plot of I_{ds} and I_g of AlGaIn/GaN HEMTs with sputtered TiN gate for $V_{ds} = 40$ V and $V_g = -7$ V.

However, as our measured devices have been passivated by PECVD Si_3N_4 , the surface hopping of electrons along the surface states is unlikely to occur as these surface states, such as dangling bonds and defects, have been eliminated by passivation process. Compared to the M-S interface states which are located at ~ 0.115 eV above the Fermi level (see Figure 5-21), the extracted E_A from AlGaIn/GaN HEMT is also well matched. Hence, the reverse gate-leakage current of HEMT device with passivation believes to be dominated by

trap-assisted tunneling mechanism. This mechanism is also in good agreement with other reports [94][234].

The temperature dependent OFF-state characteristics is measured at $V_g = -7$ V, as shown in Figure 5-24 (a). The OFF-state breakdown voltage (BV_{OFF}) is defined as the voltage at which I_{ds} reaches the current compliance of $1 \mu\text{A}/\text{mm}$ which is set to avoid the hard breakdown to the measured device. With the increase of measurement temperature, BV_{OFF} decreases gradually.

Figure 5-24 (a) OFF-state breakdown characteristics of AlGaIn/GaN HEMT at the current compliance of $1 \mu\text{A}/\text{mm}$ and (b) BV_{OFF} at $V_g = -7$ V as a function of measurement temperature

The linear fit yields a temperature coefficient of -1.75 V/K , as shown in Figure 5-24 (b). The negative or positive coefficients have been observed on AlGaIn/GaN HEMTs with much smaller values (-0.15 [236], ~ -0.13 [237],

0.33 [238] and ~ 0.05 V/K [239]). S. Arulkumaran, et al. have even reported both positive (0.28 V/K) and negative (-0.53 V/K) coefficients from an individual HEMT device at different temperature range [235]. A. P. Zhang, et al. reported the temperature coefficient of -6.0 ± 0.4 V/K for both AlGaIn and GaN SBDs and speculated that the magnitude and the sign of temperature coefficient were determined by particular defects of AlGaIn/GaN HEMT structure [240]. Compared to other reported values [236][237][238][239][240], our temperature coefficient of -1.75 V/K is within the range, suggesting that the OFF-state breakdown mechanism is dominated by the tunneling gate leakage current [235] [237].

5.7 Summary

In summary, the sputtered TiN was systematically studied as the Schottky gate on AlGaIn/GaN HEMT structure. The extracted $q\Phi_B$ value was ~ 1.1 eV by I-V and I-V-T measurements, and ~ 1.44 eV by C-V measurement, which is larger than most of the Ni/Au-free materials and other reported TiN-based Schottky contacts. The improvement of $q\Phi_B$ is likely caused by the incorporation of oxygen into sputtered TiN film, by the high $N_2/(N_2+Ar)$ gas ratio used in this work, or by the surface crystal orientation with densely packed atoms such as (111). The fabricated AlGaIn/GaN HEMTs with sputtered TiN gate exhibited good pinch-off behavior and comparable DC output and transfer characteristics to the HEMTs with Ni/Au gate. With reference to the device with

conventional Ni/Au gate, slight increase in the dynamic- R_{on} /static- R_{on} ratio was observed in the TiN gate HEMTs. As a result of the improved SBH of TiN gate, the reverse gate leakage current was reduced by one order of magnitude, and $\sim 48 - 58$ V enhanced BV_{gd} was observed. The reverse leakage current mechanisms of AlGaIn/GaN SBDs and HEMTs with sputtered TiN Schottky contacts were systematically investigated. The conduction mechanism of two terminal gate leakage current is found to be dominated by Poole-Frenkel emission rather than by Schottky emission for the reverse voltage range from 0 to -3.2 V (Reg. I). This has also been confirmed by the extraction of parameters $\epsilon_{s(h)}$ and $q\phi_t$. The obtained interface state of 0.53 eV may be due to the plasma damage to the surface of AlGaIn/GaN HEMT structure during the TiN sputtering. For the reverse voltage -20 V $< V_R < -3.2$ V (Reg. II), the gate leakage current conduction mechanism is dominated by trap-assisted tunneling with the activation energy of 0.115 eV. This is in close agreement with the activation energy (~ 0.159 eV) calculated from the three terminal OFF-state gate leakage current of AlGaIn/GaN HEMT. The observation of negative temperature coefficient (-1.75 V/K) from the OFF-state breakdown voltage ($1 \mu\text{A/mm}$) of AlGaIn/GaN HEMT is due to the trap-assisted tunneling mechanism, which is also well correlated with the conduction mechanism realized from the two terminal reverse gate leakage current of AlGaIn/GaN SBDs with sputtered TiN. Based on the above results, the sputtered TiN Schottky gate with enhanced $q\Phi_B$ is shown to be a good potential candidate for the fabrication of AlGaIn/GaN HEMT in Si CMOS compatible process line.

Chapter 6 Conclusions and Recommendations for Future Work

6.1 Conclusions

AlGaIn/GaN HEMTs have demonstrated great potentials in the high voltage and high frequency applications due to the advantages in intrinsic material properties, such as high electron saturation velocity, high sheet carrier density, and wide band gap, etc. Conventionally, AlGaIn/GaN HEMTs are realized in III-V line using the processes and materials that are not compatible to Si CMOS fabrication. For example, the mesa etching isolation will yield a non-planar wafer surface and gold-based metal contact (Ti/Al/Ni/Au ohmic contact and Ni/Au Schottky gate typically) will cause contamination issues. With the progresses in the epitaxial growth technology, crack-free AlGaIn/GaN HEMT structure can be grown on large diameter (e.g. 200 mm) Si substrate with low value of wafer bowing and low dislocation densities, which provides the possibility of fabricating devices in the existing Si CMOS line to meet the expanding market of GaN electronics. In recent years, CMOS-compatible processes have received a lot of interests as this approach allows high-volume low-cost fabrication of AlGaIn/GaN HEMTs. In addition, CMOS-compatible processes also offer the opportunity of on-chip integration of GaN devices with Si CMOS devices to enhance the functionalities of the ICs.

This thesis concentrates on the development of three key process modules

which will be CMOS compatible for AlGaN/GaN HEMTs fabrication. These are planar device isolation by ion-implantation, non-gold ohmic contact, and non-gold Schottky gate. In addition, the conduction mechanisms of metal contacts onto AlGaN/GaN HEMTs are also analyzed to understand the impact of these processes on the device characteristics.

The planar inter-device isolation is achieved using multi-energy ultra-heavy $^{131}\text{Xe}^+$ ion implantation. About an order of magnitude lower buffer leakage current, and an order of magnitude higher buffer sheet resistance and ON/OFF current ratio are observed by implanted isolation as compared to conventional mesa isolation. The breakdown voltage of implanted buffer is almost unchanged with and without Si_3N_4 passivation, which differs from the large variation in breakdown voltage for mesa isolated buffer with passivation. The extracted activation energy (0.513 eV) is attributed to the lattice damage in AlGaN/GaN HEMT structure caused by implantation with plasma. With reference to mesa-isolated AlGaN/GaN HEMTs, the devices isolated by implantation exhibit similar DC output and transfer characteristics, improved OFF-state breakdown voltages (by $\sim 106 - 128$ V, with same gate-drain space), and reduced reverse gate leakage current (by one order of magnitude). The observations of the improved performance of HEMT devices with stable buffer breakdown voltage after passivation are due to the direct contact of passivation layer or metal gate with 2DEG at the mesa side wall is eliminated by planar isolation with ion implantation.

The non-gold ohmic contact employs Ta/Si/Ti/Al/Ni/Ta metal stacks.

After annealing at 850 °C, very low contact resistance of $0.22 \pm 0.03 \text{ } \Omega \cdot \text{mm}$ is measured from TLM patterns with good repeatability and smooth surface morphology (RMS $\sim 5.5 \text{ nm}$). The increased higher annealing temperature (875 °C and 900 °C) only slightly degrades the contact resistance to $0.25 \text{ } \Omega \cdot \text{mm}$. The low contact resistance of this non-gold ohmic contact is primarily contributed by the formation of low work function Ti_xSi_y at the metal-semiconductor interface and by the intermixing of Ti_xSi_y with the bottom Ta layer, which is confirmed by XRD, HR-TEM, and EDX (line- and area-scan) analysis. As AlGaIn/GaN HEMT structure is separated by the bottom Ta from reacting with other metals in the metal stack, the current transport in this non-gold ohmic contact is dominated by thermionic emission (TE) through surmounting a low energy barrier ($\sim 0.113 \text{ eV}$ and $\sim 0.121 \text{ eV}$ for 850 °C and 900 °C annealing, respectively) at the metal-semiconductor interface, which is different from the thermionic-field emission (TFE) observed in conventional Ti/Al/Ni/Au ohmic contacts.

The sputtered TiN is evaluated as the non-gold Schottky contact for CMOS-compatible processes. The Schottky barrier height of sputtered TiN on AlGaIn/GaN HEMTs is characterized by different methods for comparison. Various techniques were carried out to extract the Schottky barrier height (SBH or $q\Phi_B$). SBH values of $1.126 \pm 0.012 \text{ eV}$ was obtained by I-V measurement, $\sim 1.098 \text{ eV}$ by activation energy plot, $\sim 1.105 \text{ eV}$ by modified Norde's function, and $\sim 1.44 \text{ eV}$ by C-V measurement. These values are larger than the conventional Ni/Au gate and most of reported non-gold materials. Compare to

other TiN-based Schottky gate on AlGaIn/GaN HEMT structure, our sputtered TiN also exhibits improved SBH which could be due to the higher N₂ ratio in the total gas flow during sputtering, or the incorporation of oxygen impurities into sputtered TiN, or the preferred surface crystal orientation (e.g. (111) surface). The mechanism study using Schottky barrier diodes reveals that the gate leakage current is dominated by Poole-Frenkel emission through an interface state of ~ 0.53 eV at small reverse bias ($-3.2 \text{ V} < V_R < 0$), and by trap-assisted tunneling via another interface state of ~ 0.115 eV at large reverse bias ($-20 \text{ V} < V_R < -3.2 \text{ V}$). The energy level of tunneling traps (~ 0.115 eV) is well agreed by the activation energy (~ 0.159 eV) calculated from three-terminal OFF-state gate leakage current of AlGaIn/GaN HEMTs. Compare to devices with conventional Ni/Au gate, the AlGaIn/GaN HEMTs with sputtered TiN gate demonstrates comparable DC output and transfer characteristics, slight increased dynamic/static ON-resistance ratio (by $\sim 8\%$), reduced gate leakage current (by one order of magnitude) and improved OFF-state breakdown voltage (by ~ 48 -58 V). These improvements are primarily attributed to the larger SBH of sputtered TiN on AlGaIn/GaN HEMTs.

6.2 Key Contributions of this work

The motivation of this thesis is to develop and investigate the transport mechanisms of three process modules namely planar isolation process by ion-implantation, non-gold ohmic contact and non-gold Schottky contact which are

the three key process modules in CMOS-compatible technology. The key contributions of this thesis include:

- An ion-implanted isolation using multi-energy ultra-heavy $^{131}\text{Xe}^+$ ions is demonstrated for the first time. Because of the heavier ion mass of $^{131}\text{Xe}^+$, more lattice damages and disorders are created hence a better inter-device isolation effect has been observed. In addition, the origin of high resistivity of implanted region is identified by analyzing the activation energy which can be used to explain the mechanism of the implant isolation.
- The non-gold ohmic contact with Ta/Si/Ti/Al/Ni/Ta metal stacks was reported with a low contact resistance ($\sim 0.22 \pm 0.03 \text{ } \Omega \cdot \text{mm}$) and excellent repeatability. Compare to other reported non-gold ohmic contacts, our contact resistance value is much lower without the need to introduce additional processes such as AlGaN barrier recess or plasma treatment prior to ohmic metal deposition. The low resistance was found to be due to the formation of low work function Ti_xSi_y at the metal-semiconductor interface. Thermionic emission (TE) mechanism was identified for the first time for such non-gold ohmic contact which was not observed in gold-based or other reported non-gold ohmic contacts.
- Sputtered TiN gate was demonstrated for both AlGaN/GaN Schottky barrier diodes and HEMTs. AlGaN/GaN HEMT structure with TiN Schottky gate metal was characterized in detailed using different methods. Larger Schottky barrier height ($\sim 1.1 \text{ eV}$ by I-V and I-V-T, $\sim$

1.44 eV by C-V) was obtained as compared to other reported TiN-based contacts (max. ~ 0.95 eV). The gate leakage current mechanism of sputtered TiN on AlGaIn/GaN HEMT structure was systematically studied for the first time. A band diagram model of metal-semiconductor interface with two different trap energy levels was used to explain the reverse leakage behavior, which has not been reported by others. The energy level (~ 0.53 eV) of interface traps was extracted and was found to be likely caused by the sputtering process of TiN.

6.3 Recommendations for Future Work

So far, the solutions for the three key challenging modules in developing the CMOS-compatible processes for AlGaIn/GaN HEMTs fabrication have been addressed in this thesis. In spite of the promising results that have been demonstrated in this work, there are some areas which can be improved. These are:

- (1) The inter-device isolation by multi-energy $^{131}\text{Xe}^+$ ion implantation exhibits improved isolation effect with lower buffer leakage and higher ON/OFF current ratio as compare to mesa isolation, which is due to more lattice damages and disorders are created by ultra-heavy $^{131}\text{Xe}^+$ ions in AlGaIn/GaN HEMT structure. However, the thermal reliability of implant isolation is not yet investigated. The lattice damages and disorders could be recovered by high temperature thermal annealing and

the buffer leakage would increase in return, indicating the degradation of isolated region. Therefore, the thermal reliability study for the maximum temperature and duration that the implant isolated AlGaIn/GaN can withstand is necessary to evaluate the capability of device operating in extreme conditions.

- (2) Further optimization could be done to improve the non-gold ohmic contact. The Ta/Si/Ti/Al/Ni/Ta non-gold metal scheme developed in this work includes six layers and five different metals, which may not be preferred in industry. Hence, one can look into reducing the number of types of metals and layers of the metal stack. The key component of the Ta/Si/Ti/Al/Ni/Ta ohmic contact is the bottom Ta/Si/Ti/Al stacks. Hence, the optimization work can focus on the Ni/Ta cap layers. In addition, the annealing temperature (850 °C) needs to be decreased to accommodate the gate-first fabrication. The AlGaIn barrier recess may be necessary to facilitate the ohmic contact formation at lower annealing temperature (target < 550 °C).
- (3) The patterns of the ohmic contact (Ta/Si/Ti/Al/Ni/Ta) and the Schottky gate (sputtered TiN) in this work are defined by metal lift-off. For the fully CMOS-compatible processes, the metal etch-back should be used instead of the lift-off. Hence, the development of metal etch-back process and the study on the impact of the metal etch-back process to device characteristics is important to the process integration. The plasma damages caused by dry etching on AlGaIn/GaN should be focused.

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List of Publications

Journal Papers:

1. **Y. Li**, G. I. Ng, S. Arulkumaran, Z. H. Liu, K. Ranjan, K. S. Ang, P. P. Murmu, and J. Kennedy, “Improved Planar Device Isolation in AlGaIn/GaN HEMTs on Si by Ultra-heavy $^{131}\text{Xe}^+$ Implantation”, *Physica Status Solidi (a)*, accepted with minor revision.
2. **Y. Li**, G. I. Ng, S. Arulkumaran, Z. H. Liu, K. Ranjan, W. C. Xing, K. S. Ang, P. P. Murmu, and J. Kennedy, “AlGaIn/GaN High electron mobility transistors on Si with sputtered TiN gate”, *Physica Status Solidi (a)*, Vol. 214, No. 3, pp. 1600555, Mar. 2017. (<http://dx.doi.org/10.1002/pssa.201600555>)
3. **Y. Li**, G. I. Ng, S. Arulkumaran, G. Ye, Z. H. Liu, K. Ranjan, and K. S. Ang, “Investigation of gate leakage current mechanism in AlGaIn/GaN high-electron mobility transistors with sputtered TiN”, *Journal of Applied Physics*, Vol. 121, No. 4, pp. 044504, Jan. 2017. (<http://dx.doi.org/10.1063/1.4974959>)
4. **Y. Li**, G. I. Ng, S. Arulkumaran, G. Ye, C. M. Manoj Kumar, M. J. Anand, Z. H. Liu, “Conduction mechanism of Non-gold Ta/Si/Ti/Al/Ni/Ta ohmic contacts in AlGaIn/GaN High-electron-mobility transistors”, *Applied Physics Express*, Vol. 8, pp. 041001, Mar. 2015. (<http://dx.doi.org/10.7567/APEX.8.041001>)
5. **Y. Li**, G. I. Ng, S. Arulkumaran, C. M. Manoj Kumar, K. S. Ang, M. J. Anand, H. Wang, R. Hofstetter and G. Ye, “Low-Contact-Resistance Non-gold Ta/Si/Ti/Al/Ni/Ta Ohmic Contacts on Undoped AlGaIn/GaN High-Electron-Mobility Transistors Grown on Silicon”, *Applied Physics Express*, Vol. 6, pp.

116501, Oct. 2013. (<http://dx.doi.org/10.7567/APEX.6.116501>)

6. G. Ye, H. Wang, S. L. G. Ng, R. Ji, S. Arulkumaran, G. I. Ng, **Y. Li**, Z. H. Liu and K. S. Ang, “Effect of surface pre-treatment on interfacial chemical bonding states of atomic layer deposited ZrO₂ on AlGa_N,” *Journal of Vacuum Science & Technology A*, Vol. 3, No. 5, pp. 05E117, Sep/Oct 2015 (<http://dx.doi.org/10.1116/1.4927164>)

7. G. Ye, H. Wang, S. L. G. Ng, R. Ji, S. Arulkumaran, G. I. Ng, **Y. Li**, Z. H. Liu, and K. S. Ang, “Impact of post-deposition annealing on interfacial chemical bonding states between AlGa_N and ZrO₂ grown by atomic layer deposition,” *Applied Physics Letters*, Vol. 109, No. 09, pp. 091603, Mar. 2015. (<http://dx.doi.org/10.1063/1.4914351>)

8. G. Ye, H. Wang, S. L. G. Ng, R. Ji, S. Arulkumaran, G. I. Ng, **Y. Li**, Z. H. Liu and K. S. Ang, “Influence of post-deposition annealing on interfacial properties between Ga_N and ZrO₂ grown by atomic layer deposition,” *Applied Physics Letters*, Vol. 105, No. 15, pp. 152104, Oct. 2014. (<http://dx.doi.org/10.1063/1.489857>)

9. G. Ye, H. Wang, S. Arulkumaran, G. I. Ng, **Y. Li**, Z. H. Liu, and K. S. Ang, “Band alignment between Ga_N and ZrO₂ formed by atomic layer deposition,” *Applied Physics Letters*, Vol. 105, No. 02, pp. 022106, Jul. 2014. (<http://dx.doi.org/10.1063/1.4890470>)

10. G. Ye, H. Wang, S. Arulkumaran, G. I. Ng, R. Hofstetter, **Y. Li**, M. J. Anand, K. S. Ang, Y. K. T. Maung, and S. C. Foo, “Atomic layer deposition of ZrO₂ as gate dielectric for AlGa_N/Ga_N metal-insulator-semiconductor high

electron mobility transistors on silicon,” *Applied Physics Letters*, Vol. 103. No. 14, pp. 142109, Oct. 2013. (<http://dx.doi.org/10.1063/1.4824445>)

Conference Papers:

1. **Y. Li**, G. I. Ng, S. Arulkumaran, K. Ranjan, Z. H. Liu, “Sputtered TiN Schottky Barrier Diodes on AlGa_N/Ga_N Heterostructures”, oral presented in 9th *International Conference on Materials for Advanced Technologies (ICMAT)*, 18-23, June 2017, Singapore
2. **Y. Li**, G. I. Ng, S. Arulkumaran, K. Ranjan, Z. H. Liu, K. S. Ang, P. P. Murmu and J. Kennedy, “Improved Device Isolation in AlGa_N/Ga_N HEMTs by Multi-energy ¹³¹Xe⁺ Implantation”, poster presented in *International Workshop on Nitride Semiconductors (IWN 2016)*, 2-7, October 2016, Orlando, Florida, U. S. A
3. **Y. Li**, G. I. Ng, S. Arulkumaran, C. M. Manoj Kumar, K. S. Ang, and Z. H. Liu, “AlGa_N/Ga_N High Electron Mobility Transistors on Si with Sputtered TiN gate”, oral presented in 43rd *International Symposium on Compound Semiconductors (ISCS) of Compound Semiconductor Week (CSW 2016)*, 26-30 June 2016, Toyama, Japan
4. **Y. Li**, G. I. Ng, S. Arulkumaran, C. M. Manoj Kumar, K. S. Ang, H. Wang, G. Ye, R. Hofstetter, M. J. Anand, “Investigations of CMOS-Compatible Non-Gold Ta/Si/Ti/Al/Ni/Ta Ohmic Contact for AlGa_N/Ga_N HEMT on Si with Low Contact Resistance”, poster presented in 10th *International Conference on*

Nitride Semiconductors (ICNS), 25-30, August 2013, Washington DC, U. S. A.

5. G. I. Ng, Y. Li, S. Arulkumaran, K. Ranjan, and J. Kennedy, “Inert-gas heavy ion implantation for device isolation in AlGa_N/Ga_N HEMTs on Si substrate”, invited talk in *20th International Conference on Ion Beam Modification of Materials (IBMM)*, 30 October – 4 November, Wellington, New Zealand,

6. G. Ye, H. Wang, S. Arulkumaran, G. I. Ng, R. Hofstetter, Y. Li, M. J. Anand, K. S. Ang, Y. K. T. Maung, S. C. Foo, "AlGa_N/Ga_N MISHEMTs on Silicon Using Atomic Layer Deposited ZrO₂ as Gate Dielectric," poster presented in *71st Device Research Conference (DRC)*, 23-26 June 2013, South Bend, Indiana, U. S. A.

7. M. J. Anand, G. I. Ng, S. Arulkumaran, H. Wang, Y. Li, S. Vicknesh, and T. Egawa, “Low Specific On-resistance and High Figure-of-Merit AlGa_N/Ga_N HEMTs on Si substrate with Non-Gold Metal Stacks”, poster presented in *71st Device Research Conference (DRC)*, 23-26 June 2013, South Bend, Indiana, U. S. A.