Improved Ni Ohmic Contact on n-Type 4H-SiC

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This paper presents the structural, chemical and electronic properties of Al/Ni/
Al-layers evaporated on 4H silicon carbide and then annealed at 1000°C for 5
min. The structure was investigated before and after annealing by transmission
electron spectroscopy from cross-sectional specimens. With x-ray photoelectron
spectroscopy, both element distribution and bonding energies were followed
during sputtering through the alloyed metal-semiconductor contact. Voids are
found in both annealed Ni/4H-SiC and Al/Ni/Al/4H-SiC contact layers, though
closer to the metal-semiconductor interface in the former case. The first alumi-
num-layer is believed to prevent voids to be formed at the interface and also to
reduce the oxide on the semiconductor surface. The contact was found to be
ohmic with a specific contact resistance $\rho_c = 1.8 \times 10^{-5}$ Ωcm^2 which is more than
three times lower ρ_c than for the ordinary Ni/4H-SiC contact prepared in the
same way.

Key words: Al/Ni/Al/4H-SiC, transmission electron microscopy, x-ray
photoelectron spectroscopy

INTRODUCTION

With the development of a wide variety of 4H-SiC
semiconductor devices, low resistance, reproducible
and stable ohmic contacts to the semiconducting
material are of great need. Since the devices based on
4H-SiC are expected to work at elevated tempera-
tures, high frequencies, and high voltages the contact
properties may even be a limiting factor for device
operation. A number of different metals have been
proposed as suitable ohmic contact materials,[1-6] and
especially Ni-based metallization systems are sug-
gested as good candidates.[7]

Nickel has been shown to react very efficiently with
SiC to form silicides,[8] unfortunately leaving unreacted
carbon atoms behind that may deteriorate the ohmic

contact. Moreover special care must be taken to avoid
oxidation during the high temperature annealing.
There have been some studies to mix nickel with
another metal, forming a stable carbide during the
annealing process, with a positive result in improving
the contact resistance.[7] Al has been reported to have
those properties we are looking for, e.g. formation of
a carbide[8] (Al_4C_3) while interface oxides are chemi-
cally reduced.[1] The aim of this study is to improve the
prospects for nickel to form good ohmic contact with
SiC by adding aluminum.

In this paper, we present a study of an Al/Ni/Al
contact on 4H-SiC before and after annealing at
1000°C for 5 min using transmission electron mi-
croscopy (TEM) and x-ray photoemission spectroscopy
(XPS). The four-point probe method developed by
Terry and Wilson and improved by Kuphal[9] was
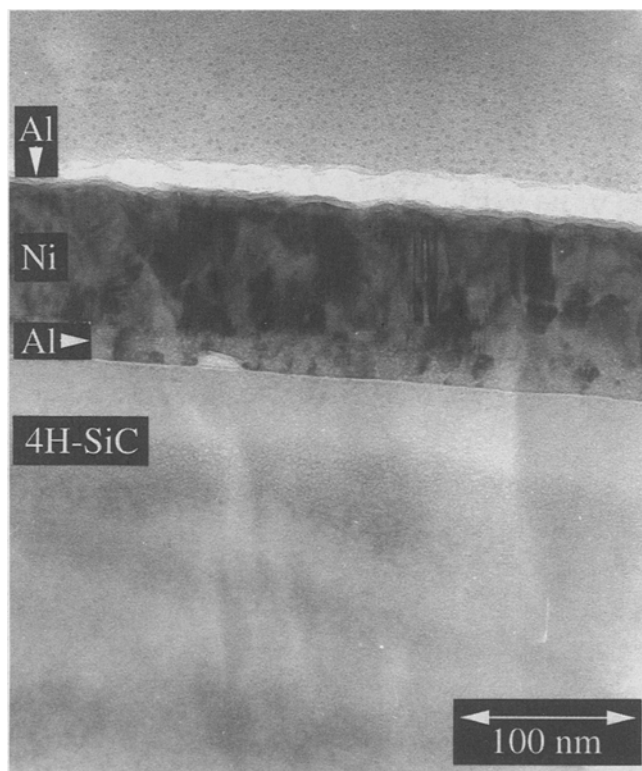employed to determine the specific contact resis-

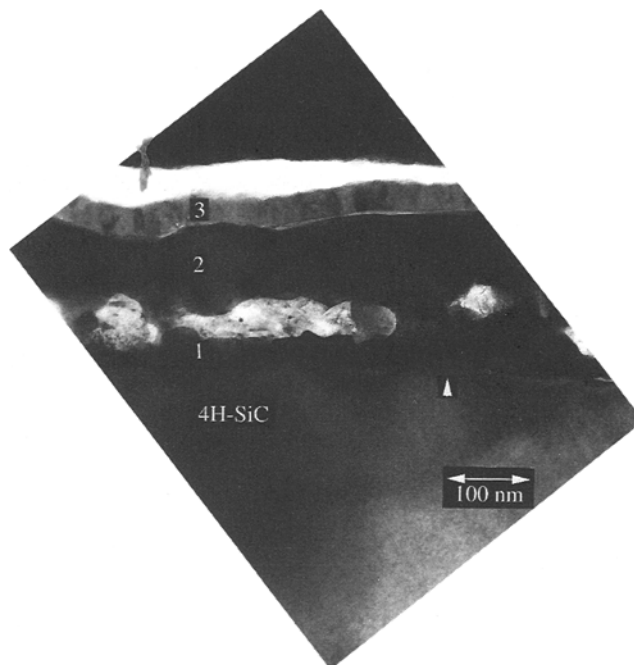Fig. 1. Bright field TEM image of an as-deposited Al/Ni/Al/4H-SiC structure.

Fig. 2. TEM image of an annealed Al/Ni/Al/4H-SiC structure, annealing temperature was 1000°C for 5 min. The three contact regions are labeled 1, 2, and 3. The arrow indicates a region of SiC which has reacted with the metals less readily than the surrounding during the annealing.

tance, ρ_c.

EXPERIMENTAL

The samples were 4H-SiC, 0.2 μm thick n-type (doping 1×10^{19} cm^{-3}) epitaxial layer grown on n-type (doping $6-7 \times 10^{18}$ cm^{-3}) off-axis $\approx 3.5°$ toward <11$\overline{2}$0> Si-face substrates.

Before metallization, the substrate surface was chemically cleaned in, $H_2SO_4+H_2O_2$, $HCl+H_2O_2+H_2O$ and $HF+H_2O$. The samples were then loaded into the evaporator system where 200Å Al, 750Å Ni, and 70Å Al were thermally evaporated at a base pressure of 2×10^{-7} mbar. Also a reference sample was prepared with a 1000Å Ni-layer. After metallization, the samples were annealed in a furnace at 1000°C for 5 min under a N_2-gas flow. The samples intended for the four-point probe measurement of ρ_c were evaporated through a shadow mask and then reactively ion etched to remove the epitaxial n$^+$ layer outside the contact pads. The distance between the contacts was 5 mm and the contact diameter was 100 μm.

XPS studies were carried out in an ESCALAB Mk II (VG Scientific) electron spectrometer.[10] The photoelectrons were excited with an Al K$_\alpha$ (1486.6 eV) x-ray source and an analyzer pass energy of 20 eV. The instrumental resolution measured as full width at half maximum (FWHM) of the Ag 3d$_{5/2}$ photoelectron peak was 1.2 eV. The C1s, O1s, Si2p, Al2p, and Ni2p photoelectron lines were recorded. The depth profiling was performed by Ar$^+$ sputtering at 3 keV and a current density of 16.08 μA/cm^2. The metal/SiC system has also been investigated in a CM 20 TEM, equipped with an energy dispersive system (EDS).

RESULTS AND DISCUSSION

TEM image of as-deposited specimen shows polycrystalline structure with total thickness $\approx$100 nm, (Fig. 1). The interface between the first Al-layer and Ni is clearly visible, while the upper Al/Ni interface is not sharp. The role of the Al top layer is to prevent Ni from oxidization. At certain places along the SiC/metal interface, there are white stripes that probably formed during the TEM sample preparation due to surface oxidation. A small amount of SiO$_x$ is also suggested by the EDS spectra measured at the first Al-layer and showing a weak O-related peak.

Figure 2 shows a TEM picture taken from an annealed specimen, the total layer thickness has increased to 180–210 nm as a result of interfacial reactions. From the figure, it is clear that reactions took place between the substrate and the metals, one can see bumps of SiC left behind intact (indicated with arrows), but the surrounding SiC has been consumed in the reaction. The layer contains a lot of voids positioned above the original location of the first Al layer. Three regions can be distinguished in the annealed contact: the first region is in contact with the SiC, the second region which contains relatively large grains is above the voids and the third region is a covering layer of the metallization which is about 30 nm thick. From the EDS analysis, one can conclude that: the first region does not contain any Al, although Al was deposited first, the second region contains some Si beside the Ni, and the third region is the only

region where Al and oxygen contents are significant.

Figure 3 shows the element distribution in the alloyed Al/Ni/Al/4H-SiC structure after depth profiling. Since the film thickness varies, the upper surface is rough and the film contains large voids, special care must be taken when interpreting the XPS depth profiling. Within the range of 0–50 min sputtering, the percentage of O, Al, and Ni is 60, 37, and 2 at%, respectively, while Si and C are not observed. With increasing sputtering time, the percentages of O and Al decrease and after 115 min they are equal to zero. Within the 50–115 min interval, the percentages of Si and Ni increase and reach maximum values of about 28 and 64 at%, respectively. Further sputtering results in a drop of the Si and Ni percentages to about 8 and 13 at%, respectively, in this same interval C concentration rises and attains about 80 at% at 200 min. After 225 min sputtering, the concentration of C sharply decreases, and after 325 min both C and Si are almost equal to 50 at%.

When following the photoelectron peak maximum shift with increasing sputtering time, Fig. 4a, and compare with binding energy (BE) for different compounds, one can see that Al is first in the oxidized state (BE of 74.6 eV). The peak maximum then shifts toward 75.6 eV which corresponds to Al bonded to carbon (Al_4C_3).

After 65 min sputtering, two weak Si2p peaks are observed (Fig. 4b): one with a maximum at 99.3 eV and the other at 103.2 eV which are characteristic of Si in a silicide and a oxide form, respectively. After 95 min sputtering, the oxide peak of Si disappears. Within the 210–280 min interval, the Si2p peak maximum is at 99.8 eV, probably due to overlapping peaks of Si in nickel silicide (BE of 99.5 eV) and Si in SiC (BE of 100 eV), respectively. Peak maximum is at 100 eV after 350 min.

Ni (Fig. 4c) is mainly in the oxide state (BE of 855.3 eV) in the beginning, while after 85 min it is bonded to Si forming nickel silicides (BE of 853.2 eV). It is probable from Fig. 3 that after a sputtering time of 115 min when the Ni and Si concentrations reach their maximum, the nickel silicide has the stoichiometry of Ni_2Si; whereas after 200 min sputtering, the nickel silicide has the stoichiometry of NiSi.

The C1s peak maximum (Fig. 4d) at the surface is at 285 eV and is associated with carbon from adsorbed hydrocarbons. Within the 65–100 min interval the C1s peak has its maximum at 283 eV, which is characteristic of carbon in Al_4C_3. After that, up to 200 min sputtering the C1s peak has its maximum at 284.5 eV, which is typical of carbon in the graphite state. For longer sputtering time the peak shifts toward 283.2 eV which represents carbon in SiC.

The specific contact resistance was found to be $\rho_c = 1.8 \times 10^{-5}\ \Omega cm^2$, which is more than three times lower than for the Ni contact ($\rho_c = 6 \times 10^{-5}\ \Omega cm^2$) using the same equipment and procedure (see also Ref. 11).

When comparing Fig. 2 with the element distribution results from the XPS, we can see that they really complete each other. The top layer is not flat,

from hill-top to valley it is almost 30 nm at some parts and this reflects in the XPS spectra (slow distribution changes). The first 50 min of sputtering indicate Al_2O_3 which is consistent with the ≈30 nm thick top layer. After 50 min of sputtering, we enter the second region where the Al_2O_3-related peak goes down and it is replaced by Al_4C_3, nickel silicide and a small quantity of SiO_x. Next sputtering period consists of an increasing silicide-related peak followed by an increasing carbon peak. In the TEM-image, one can see voids formed approximately at the same depth where the carbon concentration increase is observed. This sug-
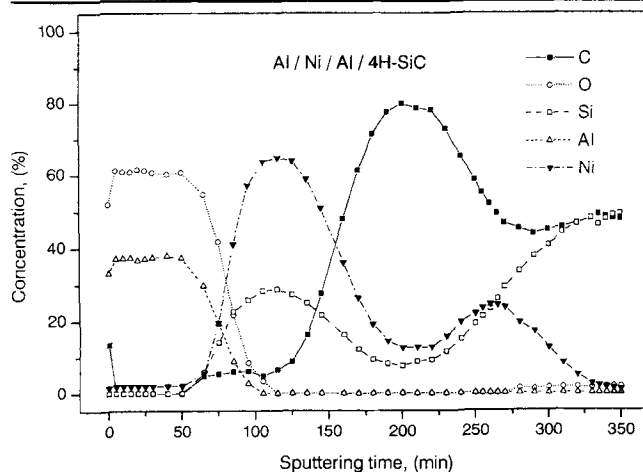

Fig. 3. X-ray photoemission profiles of an Al/Ni/Al/4H-SiC structure annealed at 1000°C for 5 min.
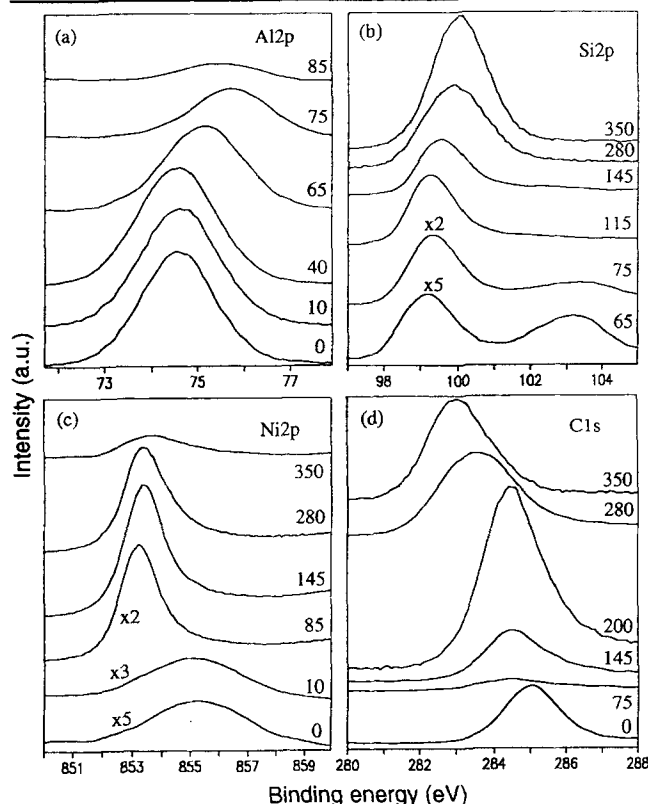

Fig. 4. Photoelectron peak of the annealed Al/Ni/Al/4H-SiC structure at different sputtering times (in minutes): (a) Al2p, (b) Si2p, (c) Ni2p, and (d) C1s.

gests that most of the carbon left after the silicide formation has partly filled up the voids. Continuing sputtering shows an increase of silicide and SiC and a decrease of carbon in graphite state, thus we have entered the third region and the SiC substrate at some places.

To summarize, due to interdiffusion processes within the metal layers, we can assume the following picture for the metal/SiC structure. The upper Al layer is mainly consumed by forming Al_2O_3. The Al that has not reacted with oxygen or carbon at the SiC interface,[1] can diffuse upward the surface, but the strong Ni-Si chemical affinity[3] gives rise to a faster diffusion of Ni toward SiC. Due to this, an excess of vacancies is produced which precipitate as voids, so called Kirkendall porosity.[12] The voids are normally found above the original Ni-Al interface. We observed the same porosity when annealing Ni/SiC, but the porous region is in the vicinity of the reacted interface and the nickel-silicide layer itself also contain a lot of small voids. Thus, the Al does not participate in creating the actual ohmic contact, but it reduces the interface oxide, reacts with carbon and also extends the distance for the fast diffusing Ni. This may be the reason why we observe lower ρ_c in the Al/Ni/Al/SiC case compared with Ni/SiC.

CONCLUSION

When depositing Al/Ni/Al on 4H-SiC instead of only Ni an improvement of the specific contact resistance can be achieved. The first deposited aluminum layer is believed to prevent void formation at the interface, reduce the semiconductor oxide and to react with some of the carbon left behind during silicide formation.

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