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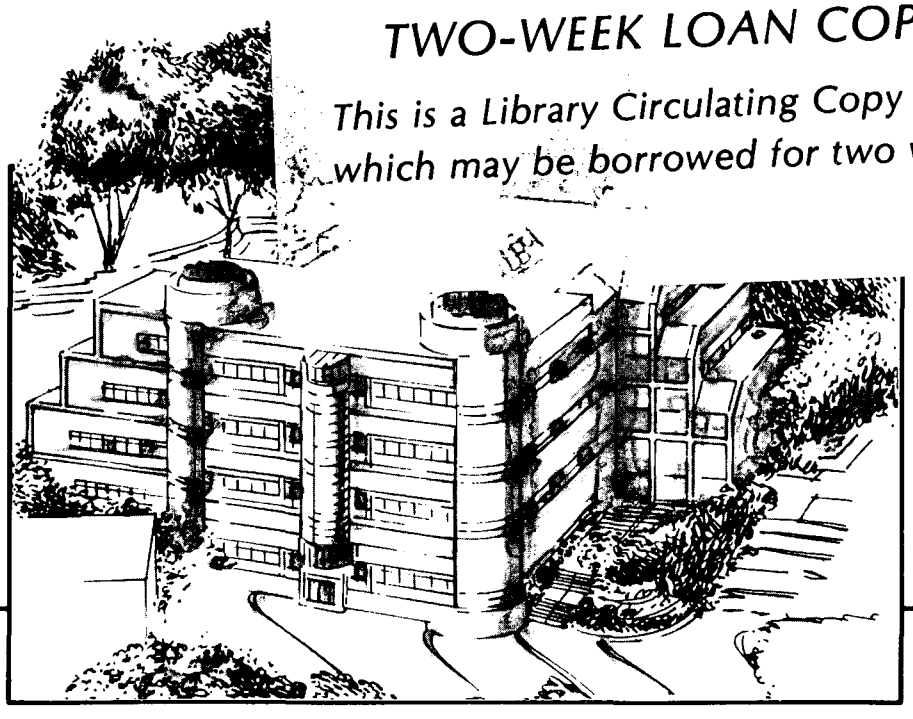

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NEW Al-Ni-Ge CONTACTS ON GaAs: THEIR STRUCTURE AND ELECTRICAL PROPERTIES

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ABSTRACT

The structure and composition of the recently developed Al-Ni-Ge ohmic contacts to n-GaAs were investigated by transmission electron microscopy combined with secondary ion mass spectroscopy (SIMS) and Auger spectroscopy. The semiconductor/metal-alloy interface of these contacts remain very flat after annealing (500°C, for 1 min - contact resistance $1.4 \times 10^{-6} \Omega cm^2$), in contrast to the widely used Au-Ni-Ge contacts. The metal sequence during deposition is found to be a critical factor in determining the electrical contact properties and the dispersion of the oxide layer on the semiconductor surface after chemical cleaning. Ge doping of the GaAs beneath the contact layer was observed by SIMS, and a tunneling mechanism through the n^+GaAs:Ge layer was proposed to explain the ohmic properties of the contacts.

1. INTRODUCTION

The properties of the ohmic contact influence the performance and reliability of semiconductor devices. The semiconductor GaAs have wide applications in microwave and opto-electronic areas as well as in fast integrated devices for data processing. One of the most important, yet still not well understood, areas of research for III-V semiconductors is the production of stable, reproducible ohmic contacts with low contact resistance. In the last few years, a number of papers have focused on this topic [1-4]. Most of them are concerned with the widely used Au-Ge-Ni/GaAs contact system. The mechanism for the observed electrical characteristics still lacks a fundamental understanding because of its complex composition, nonuniform interface, and only partly controllable electrical performance. In this paper the structure and electrical characteristics of Al-Ni-Ge ohmic contacts will be discussed. Zuleeg [5] reported that these new contacts have several advantages over Au-Ni-Ge contacts due to the higher eutectic temperature of Al-Ge 424°C [6] over AuGe [6]. These contacts show improved thermal stability compared to Au-Ni-Ge. The new contacts have a higher radiation tolerance, because of the lower atomic number of Al compared to Au [5]. In addition, evaporated Al interconnects have lower sheet resistivity than sputtered Au interconnects. It has been shown that Al-Ni-Ge can make ohmic contacts to both n- and p-GaAs [7,8].

II. CONTACT FABRICATION

Two types of Al-Ni-Ge contacts, which differed only in their layer-sequence, were investigated. One type of contact was prepared by evaporation of 30 nm of Ge on GaAs following by evaporation of 30 nm of Ni and 200 nm of Al. (This type of contact is shown schematically in Fig. 1a and will be referred to in this paper as "A"). Semi-insulating liquid-encapsulated Czochralski (LEC) GaAs was used as the substrate, which was implanted with Si^+ to form an n-type material under the contacts. A typical peak impurity concentration was in the range of 10^{18} cm^{-3}. Rapid thermal annealing at 825°C was used to activate the ion implantation. The metal was evaporated onto a room-temperature substrate in the sequence described above. After annealing at 500°C for 1 min in forming gas, a 1.4×10^{-6} Ωcm^2 contact resistance was obtained (Fig. 2). This is comparable to the resistance of Au-Ni-Ge contacts, where the typical contact resistances are found to be in the range of $2\text{-}7 \times 10^{-6}$ Ωcm^2 [3,4]. To determine if the layer order plays an important role, a second set of contacts was prepared where a 5-nm-thick Ni layer was deposited on semi-insulating GaAs, followed by an Al-Ge layer 100 nm thick (prepared in eutectic composition), a Ni layer 30 nm thick, and an Al layer 50 nm thick. (This type of contact is shown schematically in Fig. 1b and will be referred to in the paper as "B"). The samples did not receive any additional cleaning after manufactured wafer surface finish. All elements have been evaporated from Al_2O_3 crucible and out diffusion of oxygen from the crucible was detected by Auger sputtered depth profile (Fig. 3). The samples were annealed at 500°C for 3.5 min. For the "B" samples it was not possible to measure the contact resistance because of the semi-insulating substrate, but similar structures deposited on n-type GaAs (prepared by Si^+ ion implantation) did not show ohmic behavior [9]. Therefore, it was assumed that these contacts do not produce ohmic behavior, but it was of interest to investigate the structural differences between the two types of contacts.

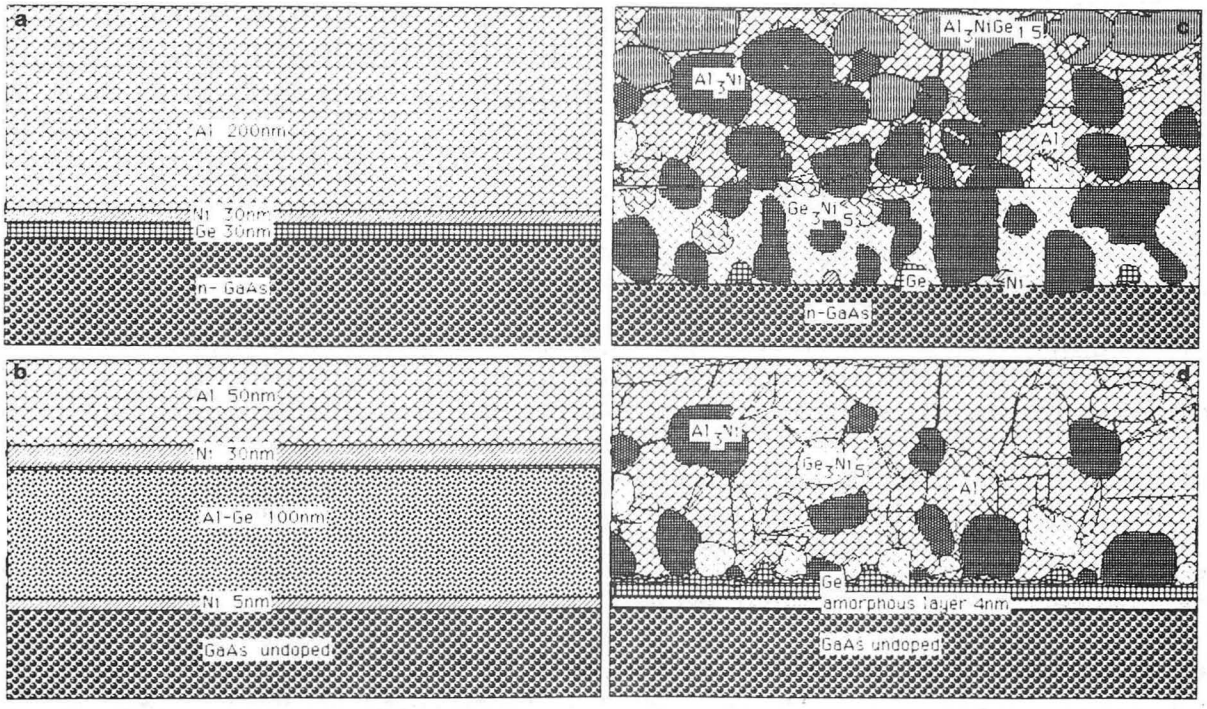

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Fig. 1. Schematic drawings of metal sequence of as-deposited and annealed Al-Ni-Ge contacts on GaAs: a) as-deposited "A" contact; b) as-deposited "B" contact; c) annealed "A" contact; d) annealed "B" contact.

III. EXPERIMENTAL METHODS

The study was carried out using transmission electron microscopy, microdiffraction, convergent-beam electron diffraction (CBED), bright- and dark-field high-resolution imaging, and a microanalytical method - energy dispersive x-ray Spectroscopy (EDX). The JEOL 200CX electron microscope, with point-to-point resolution of 2.4 Å, and the Atomic Resolution Microscope at LBL, with point-to-point resolution of 1.7 Å, were used for investigating the structure of the contacts; a JEOL 200CX analytical TEM/STEM equipped with Kevex high-angle and ultrathin-window x-ray detectors was used for analytical study. In addition, secondary ion mass spectroscopy (SIMS) and Auger sputtered depth profile were used to check the layer composition before and after annealing. Cross-section samples were prepared by gluing two samples metal-layer to metal-layer and by mechanical polishing to achieve a thickness of ~50 μm. Silver epoxy was used to glue the samples. An argon ion mill with a low-energy gun and a cold stage was used to obtain electron transparency of the samples.

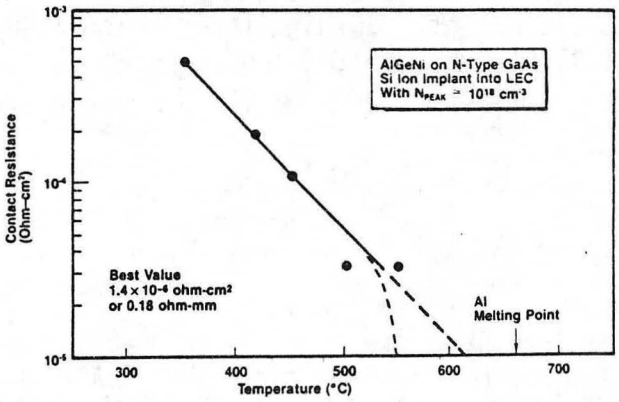

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Fig. 2. Contact resistance dependence of Al-Ni-Ge "A" contacts on alloying temperature.

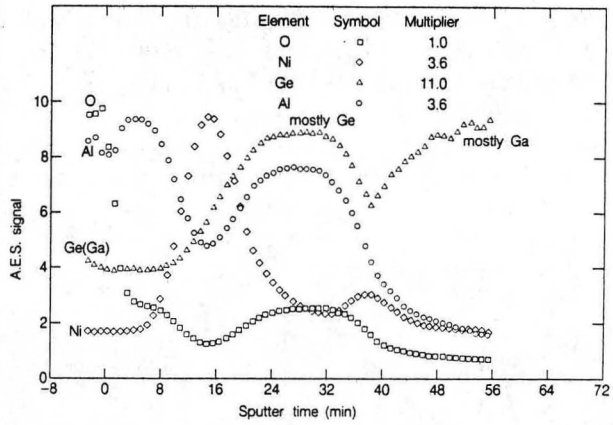

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Fig. 3. Auger depth profile of metal element distribution for as deposited "B" contacts. Due to Ge-Ga peak overlap a factor analysis routine was used to separate these two peaks [11,12].

IV. UNALLOYED CONTACTS

The as-deposited "A" contacts showed a slightly undulating interface with GaAs. Below the contact layer dislocation loops were formed in the GaAs up to 300 nm from the interface as the effect of ion implantation, with the highest density ~200 nm from the interface. In most of the investigated contact areas a layer of Ge_3Ni_5 ~17 nm thick was found in intimate contact with the GaAs (Figs. 4a,b). There were a few areas where this compound was not found, and in these areas an amorphous oxide layer was present at the interface. It is not clear if Ge_3Ni_5 was formed during metal deposition or if it was formed during TEM sample preparation, which required 30 min of heating at 90°C to allow hardening of the silver epoxy used to prepare the cross-section samples. SIMS data (Fig. 5a) show a slight shift in the Ni peak toward the Al layer, which would suggest that not all the deposited Ni was consumed in forming a new phase. Indeed, Fig. 4 shows the first layer ~17 nm thick and the second layer of Ni, with some imbedded grains of Ge, ~38 nm thick, which as a whole would make the deposited layer 55 nm thick, comparable to the expected 60 nm-thick layers of Ge and Ni.

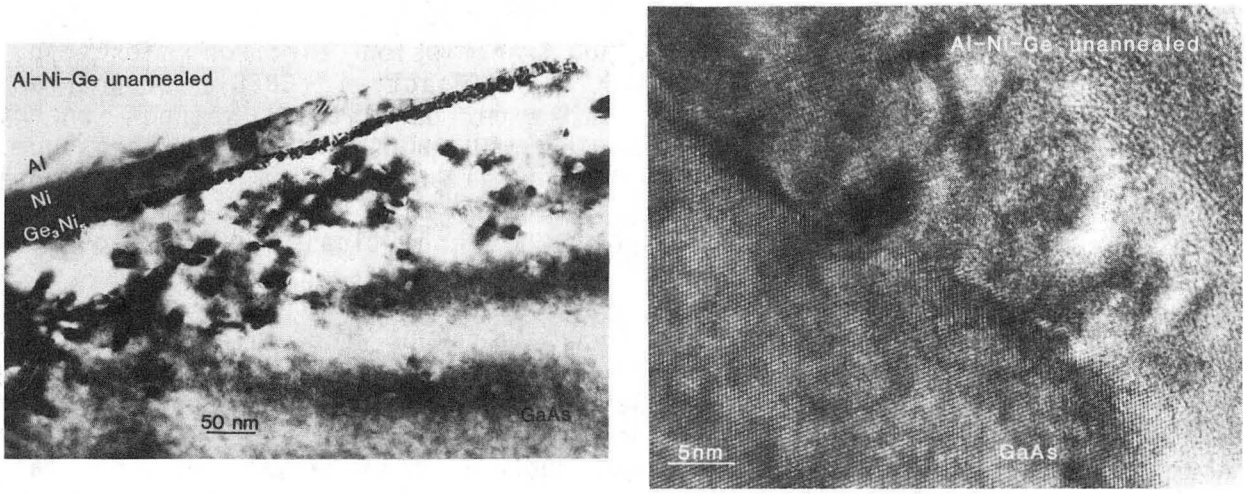

Fig. 4. TEM micrographs of the cross sections of as-deposited "A" contacts: a) low magnification of the layer sequence, b) Ge_3Ni_5 layer formed in intimate contact with GaAs. (a, XBB 884-3995 ; b, XBB 870-9370)

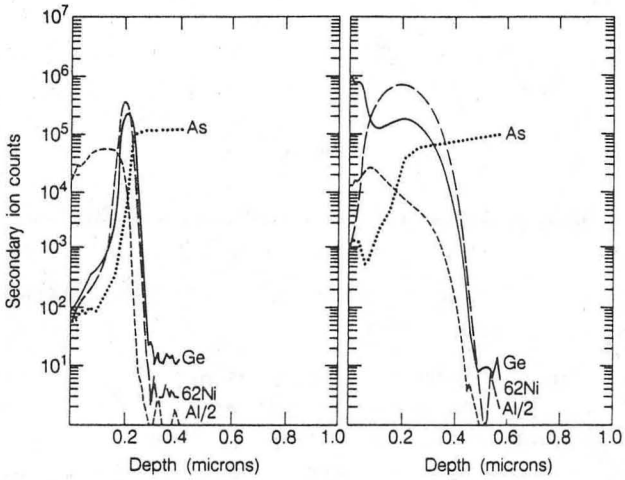

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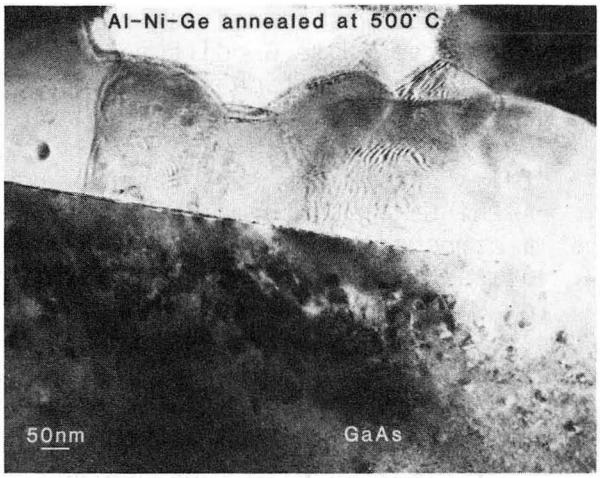
Fig. 5. SIMS spectra of metal ele-element distribution for "A" contacts: a) as deposited, b)annealed at 500°C for 1 min. (Courtesy of C. J. Hitzman, Charles Evans Lab.)

The as-deposited "B" contacts showed a layer of oxide at the interface in all cases; however, it was not clear from the TEM investigation if the first layer of Ni was physically separated from Al-Ge eutectic. However, Auger data suggests that the reaction did not take place in this case (Fig. 3). For these samples the interface with GaAs was very flat (even flatter than for the "A" samples).

V. ALLOYED CONTACTS

The annealed "A" contacts showed an average increase of 15-20% thicknesses of the metal overlayer and very flat interfaces with undulations no larger than 1.0-1.5 nm. After annealing, the interface was more uniform than the interface of the as-deposited samples (Fig. 6a,b). There were only a few areas where separated amorphous islands were present at the interface. In all other cases alloyed metals were in intimate contact with GaAs (Fig. 7). Two major phases were present in direct contact with GaAs; Ge_3Ni_5 and Al_3Ni. This shows that the Ge_3Ni_5 phase was dispersed during annealing and that Al was diffusing toward the GaAs substrate.



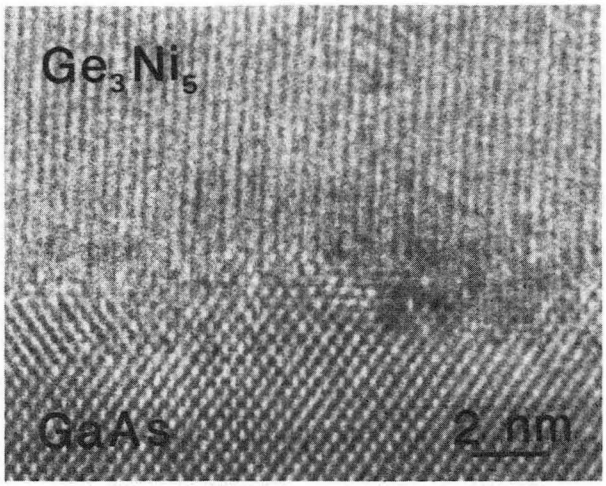


Fig. 6. a) TEM micrograph of cross section of Al-Ni-Ge "A" contact after annealing at 500°C for 1 min, b) high-resolution micrograph of the same interface. (a, XBB 870-9368; b, XBB 884-3354)

A new hexagonal ternary Al-Ni-Ge phase was found on top of the alloyed layer. Based on EDX data, the composition of this phase can be tentatively determined as $Al_6Ni_2Ge_3$. In some areas islands of another phase ~3 nm long and ~1 nm thick were found at the interface. Only one set of 0.2-nm lattice spacings was observed in these islands, not enough data to unambiguously determine the composition. A schematic distribution of the newly formed phases is shown in Fig. 1c.

Both species, Ga and As were found by EDX in the alloyed layer, with ~15% more Ga than As (using GaAs as compositional standard). Therefore, even if some scattering from the GaAs is present in the layer, the different Ga/As ratios in the overlayer and in the substrate suggest the presence of both elements in the overlayer (or at least the presence of Ga). However, new phases that contain Ga and As were not found during the TEM study. This result suggests that the presence of these elements are due to solid solution or that new ternary or quaternary phases are formed with lattice parameters close to Ge_3Ni_5 and Al_3Ni.

The annealed "B" contacts did not show an increase in alloy layer thickness. In all cases a ~4nm-thick amorphous layer was observed. For the "B" contacts the interface with GaAs was very flat, with no undulation larger than 0.3-0.5 nm. A Ge layer 7-15 nm thick was observed above the amorphous layer, and a layer consisting mostly of Al with imbedded Al_3Ni and Ge_3Ni_5 grains was present above the Ge layer (Fig. 7). A schematic distribution of the newly formed phases is shown in Fig. 1d.

Both species Ga and As were detected by EDX in the alloyed layer in the samples "B", with the Ga/As ratio identical to the ratio observed in the substrate. This result could indicate that observed Ga and As were coming from scattering from the GaAs substrate. However, these two elements were observed in Auger spectra as well. Therefore, it is probable that both species are dissolved in the alloyed layer, with a ratio similar to that in GaAs.

VI. DISCUSSION AND CONCLUSIONS

This study shows that electrical contact properties are related not only to the particular elements present in the alloyed contacts but also to the sequence in which they are deposited. It appears that Ge deposited as the first layer, in contacts referred to as "A" in this paper, partially remains at the interface, forming a compound with Ni (Ge_3Ni_5). This

Fig. 7. High-resolution micrograph of "B" contact after annealing at 500°C for 3.5 min. Note the amorphous oxide layer on the interface with GaAs and the Ge layer formed above it. The Al layer with embedded Al_3Ni grains, Ge_3Ni_5 grains and Ge grains is present in the upper part of the layer above Ge. (XBB 883-2498)

compound was formed at room temperature. Some Ge diffuses toward the contact surface layer, forming a hexagonal ternary phase with Al and Ni. Based on EDX analysis, this phase can be described as $Al_6Ni_2Ge_3$. The remaining Ni forms a compound with Al, and in many places the Al_3Ni phase was found on the interface as well as in the upper part of the layer. The phase formation after alloying is shown schematically in Fig. 1c. The interface with GaAs remains very flat, despite the new phases formed at the interface. Only in a few cases islands of the semiconductor native-oxide were found at the interface. Similar compounds were reported [8] to be formed in alloyed Al-Ni-Ge contacts deposited on p-GaAs (Zn doped, $N_A \sim 10^{19} cm^{-3}$), e.g., Al_3Ni and NiGe close to the interface with GaAs, and Ge-rich grains in the upper part of the layer consisting Al, Ni and Ge with the ratio 55:8:37. In our case this phase appears to have the ratio 6:2:3. Slight differences in the composition can be associated with some discrepancy in EDX measurements due to either the sample thickness, scattering from the surrounding grains, or even different detector sensitivities. It may be that this phase is not stoichiometric and can exist with different concentrations of particular elements. The presence of Ge in the GaAs was detected by SIMS (Fig. 5b) in the "A" contacts which did not show any continuous layer of oxide on the interface with GaAs. However, Ge was not detected in the same samples by EDX. This would suggest that the concentration of Ge in the GaAs beneath the contact is distributed very uniformly without inclusions and is below 1% concentration, the limit for detection by EDX. For the as-deposited "A" contacts, Ge was not detected by either EDX or SIMS, and in this case the electrical properties of such contacts were not ohmic. A careful study of Ge-Pd contacts on GaAs was done by Palmstrøm et al. [10] using backside SIMS. By using this method it was possible to determine the consumption of GaAs and the Ge doping concentration beneath the contact to a depth resolution better than 5 nm. Their results show that SIMS can be a reliable method for determining the concentration and layer depth through which Ge can penetrate.

The contact structure appears to be different when the layers were deposited in another sequence (referred to as "B" contacts in this paper). The Ni layer deposited as the first layer, followed by an Al-Ge eutectic, did not disperse the oxide layer present on the semiconductor surface. The monocrystalline Ge layer was formed above a ~4 nm thick oxide layer. Ge probably out-diffused from the Al-Ge eutectic composition without forming a eutectic. Ni diffuses toward the contact surface, forming an Al_3Ni phase, which in many cases was found to be imbedded in the Al. The presence of the oxide layer was the barrier to Ge diffusion into the GaAs beneath the contact.

This study suggests that Ge is probably responsible for the ohmic behavior of Al-Ge-Ni contacts. The higher concentration of Ga observed in the alloyed "A" contact suggests that Ga vacancies were formed in GaAs during annealing and were substituted by Ge, forming an n^+ layer in the

GaAs. This interpretation is consistent with Zuleeg's predictions [5] that a tunneling mechanism through the n^+ layer is responsible for the ohmic behavior of these contacts. Amphoteric behavior of Ge substituting As sublattice would explain the ohmic properties of Al-Ni-Ge contacts on p-GaAs. The fact that Ge couldn't be detected by EDX (in our study and Graham's [8] study) does not exclude the presence of Ge in the GaAs. Such small Ge concentrations can be detected easily by SIMS. More extensive studies are necessary to explore these very promising new ohmic contacts, which have very flat interfaces even after annealing and which can be applied to both n- and p-GaAs. It is possible that the lack of phases that can be formed with Ga and As are responsible for this very uniform interface. Exploring the proper layer sequence and annealing conditions could lead to achieving even lower resistance for Al-Ni-Ge contacts.

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