Double-Sided Nonalloyed Ohmic Contacts to Si-doped GaAs for Plasmoelectronic Devices

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ABSTRACT: There is increasing demand for the ability to form ohmic contacts without lossy intermediate layers on both the top and bottom sides of metal–semiconductor–metal plasmoelectronic devices such as quantum cascade lasers and metasurface photodetectors. Although highly Si-doped n-GaAs surfaces can allow an ohmic contact without alloying, conditions for realizing nonalloyed ohmic contacts to other n-GaAs surfaces, originally buried inside but exposed by removing the substrate, have yet to be studied. We discovered that nonalloyed ohmic contacts to initially buried surfaces with a practically low contact resistivity down to 77 K can be realized by fulfilling certain requirements, specifically keeping the Si-doping concentration within a narrow range of $7.5 \times 10^{18}$ to $1.25 \times 10^{19}$ cm$^{-3}$ and setting the growth temperature of the succeeding upper layers to a low value of 530 °C.

INTRODUCTION

Metal–insulator–metal structures that strongly confine light between two metal–insulator interfaces in the form of surface plasmon polaritons are one of the most fundamental architectures in plasmonics, metamaterials, and nanophotonics.1–3 They have extensively been applied to nanofocusing,4 enhanced spectroscopy,5 infrared emitters,6 and lasers.7 Although simple insulators or fluorescent films have been inserted between metals in the conventional studies, there is growing interest in metal–semiconductor–metal plasmoelectronic devices: new generation structures sandwiching epitaxially grown semiconductor quantum optoelectronic devices between metal layers. Metal–metal (MM)-waveguide quantum cascade lasers (QCLs) with enhanced mode confinement can be viewed as one of such devices (Figure 1a).8–10 Quite recently, striking performance of metasurface quantum-well infrared photodetectors (QWIPs) was also reported (Figure 1b).11–13 Many of them are based on Au-GaAs/AlGaAs-Au configurations. In these devices, the metal–semiconductor interfaces need to achieve electrical functionality for current flow with low contact resistance in addition to showing optical functionalities to be able to efficiently guide and confine the electromagnetic fields. To attain optimal optical functionalities, the formation of well-defined and abrupt interfaces between the semiconductor and metals is necessary on both sides. When alloyed intermediate layers are formed to create ohmic contacts, uncontrolled micro- and/or nanosized alloyed structures with random compositions significantly degrade the plasmonic properties of metal–semiconductor interfaces.14,15 To solve this problem, the ohmic contacts on both sides of these devices should be obtained using nonalloying methods. The separation between the metal layers (i.e., the thickness of the semiconductor) should also be as thin as possible to enhance the electromagnetic field, which

Figure 1. Schematic illustrations of metal–semiconductor–metal plasmoelectronic devices. (a) MM-waveguide QCL and (b) metasurface QWIP. (c) Standard fabrication sequence for these devices based on (c-1–c-5) substrate inversion process. The red dashed lines indicate the surfaces of the initial side contacts studied in this work.
requires thin contact layers. Finally, the necessary ohmic characteristics should be maintained even at low temperatures because metal–semiconductor–metal plasmoelectronic devices are often operated under cryogenic conditions.6–13 To fulfill these emerging requirements for standard n-GaAs-based devices, it is necessary to revisit the classical challenges of the formation of ohmic contacts to Si-doped n-GaAs. As is well known, ohmic contacts, which exhibit a linear current–voltage (I–V) characteristic, are one of the most important elemental technologies for semiconductor optoelectronic devices. Since 1970s, there has been considerable research carried out on ohmic contacts to the standard Si-doped n-GaAs (100).14–20 Because of the compensation effects of Si, the maximum active carrier concentration in the Si-doped GaAs is usually limited up to the carrier concentration \( N_C \) of \( 5 \times 10^{18} \text{ cm}^{-3} \).21–23 Because this carrier concentration is insufficient for the formation of nonalloyed ohmic contacts, alloyed ohmic contacts have normally been used.14–20 Ohmic contacts with low contact resistivity are realized by depositing metal (+ dopant) films such as AuGe or InSn onto the n-GaAs, followed by annealing at around 400 °C. The process produces alloyed intermediate layers between metals and semiconductors, which are not acceptable for plasmoelectronic devices as mentioned above.

Metal–semiconductor–metal plasmoelectronic devices are normally fabricated using the following sequence based on the substrate inversion process, as schematically shown in Figure 1c: growth of an initial-side contact layer (“initial-side” means that the contact layers are grown at the initial stage of growth) on the sacrificial layer, device layer (such as QCL and QWIP), and end-side contact layer (“end-side” means that the contact layers are grown at the end of the growth) (Figure 1c-1), metal-electrode formation on the end-side contact layer (Figure 1c-2), bonding the original substrate to the support substrate upside down using Au–In–Au10–12 or Au–Au12,13 bonding (Figure 1c-3), etching the original substrate and sacrificial layer (Figure 1c-4), and metal-electrode formation on the initial-side contact layer (Figure 1c-5). Between these two sides of the contact formation (Figure 1c-2, c-5), the way for making a nonalloyed ohmic contact with the end sides (Figure 1c-2) is straightforward. According to the previous studies, nonalloyed ohmic contacts with low contact resistivity can be realized by using (i) in situ metallization,24 (ii) doping of ultrahigh-concentration Si,3,25 or (iii) capping with low-temperature-grown GaAs (LT-GaAs) layers. Thus, if we can also achieve ohmic contacts with the initial sides (as indicated by the red dashed line in Figure 1), double-sided nonalloyed ohmic contacts will become feasible. However, nonalloyed ohmic contacts to the initial sides have not yet been established. In recent reports on metal–semiconductor–metal devices, the initial sides are nonalloyed Schottky or alloyed shallow ohmic materials using Pd as a compromise.6–10,12,13,19,20 There are several restrictions on the realization of nonalloyed ohmic contacts to the initial sides. The initial-side surfaces are first burried by the crystal growth of the succeeding upper layers and then exposed after bonding to the support substrates upside down, followed by etching of the original substrates and sacrificial layers. This rules out the in situ metallization technique ((i) above).24 It is also necessary to grow high-quality epitaxial layers such as QCL or QWIP on the initial-side contact layers. Therefore, LT-GaAs with a low-crystal quality ((iii) above) cannot be used.26,27

In this paper, we report a way of realizing nonalloyed ohmic contacts to the initial-side n-GaAs layers by utilizing the technique of doping of ultrahigh-concentration Si ((ii) above).23 We found that nonalloyed ohmic contacts cannot be realized by simply applying the technique using ultrahigh-concentration Si, which is effective for the end sides.23 By carefully optimizing the Si-doping concentration and growth temperature, nonalloyed ohmic contacts to the initial sides were, for the first time, realized in n-GaAs-based devices.

## RESULTS AND DISCUSSION

We began by examining whether the technique using ultrahigh-concentration Si, which is effective for the end sides,23 is also applicable to the initial sides. We grew a device layer (the QWIP structure consists of Al\(_{0.33}\)Ga\(_{0.67}\)As with n-GaAs QWs) sandwiched by initial-side and end-side contacts on an AlGaAs sacrificial layer. The initial-side contact layer consists of (1) 30 nm n-GaAs (the Si concentration \( N_Si = 5.5 \times 10^{19} \text{ cm}^{-3} \)), (2) 15 nm n-GaAs (\( N_Si = 3 \times 10^{18} \text{ cm}^{-3} \)), and (3) 5 nm GaAs (nondoped), which were grown on an Al\(_{0.55}\)Ga\(_{0.45}\)As sacrificial layer at 530 °C. After the growth of the initial-side contact layer, the sample was heated to 880 °C, and the device layer was grown at the same temperature. Finally, the end-side contact layer, which consists of (3) 5 nm GaAs (nondoped), (2) 15 nm n-GaAs (\( N_Si = 3 \times 10^{18} \text{ cm}^{-3} \)), and (1) 30 nm n-GaAs (\( N_Si = 5.5 \times 10^{19} \text{ cm}^{-3} \)), which were grown at 530 °C. Immediately after the growth of the end-side contact layer, the substrate was cooled down to room temperature. The high concentration of Si (\( N_Si = 5.5 \times 10^{19} \text{ cm}^{-3} \)) in the layer (1) is realized by \( \delta \)-doping of Si (the \( \delta \)-doping density \( N_{2DSi} \) is \( 1 \times 10^{13} \text{ cm}^{-2} \)) in n-GaAs (\( N_Si = 5 \times 10^{18} \text{ cm}^{-3} \)) with a 2 nm separation. The number of \( \delta \)-doping (\( M_\delta \)) is 16 because \( \delta \)-
doping was also performed on top of the sacrificial layer. After Ti/Au electrode formation, we first evaluated the contact properties of the end side. After the formation of the Ti/Au electrodes, the end-side contacts exhibited linear $I$–$V$ curves with a contact resistivity ($\rho_c$) of $3.9 \times 10^{-4} \, \Omega \cdot \text{cm}^2$ at 77 K, as shown in Figure 2a. This result is consistent with ref 25. Next, we evaluated the contact properties of the initial side at 77 K after the substrate inversion process. For this sample, we bonded the original substrate to the support substrate using Au–Au diffusion, as shown in Figure 1c. Figure 2b shows the $I$–$V$ characteristic between the two electrodes on the initial side at 77 K. Nonlinearity is clearly visible, indicating the formation of Schottky-like contacts. Because the contact resistance is too high, we cannot estimate the contact resistivity by transmission line model (TLM) measurements for this sample. It is thus concluded that the reported method for the end sides is not applicable to the initial side. A new approach is necessary.

On the n-GaAs surface, the Fermi level is pinned by the surface states located in the middle of the bandgap, which causes the formation of surface-depleted region.25,28 To reduce the thickness of the surface depletion region for efficient electron tunneling and improve contact resistivity, it is necessary to increase the active carrier concentration. Here, we calculate the contact resistivity using a simple model in which the ionized donors are uniformly distributed, and the surface Fermi level of n-GaAs is pinned by the surface states located in the middle of the bandgap (inset of Figure 3). The

$$\rho_c = \frac{1}{\frac{m_e q^2}{2\pi^2 \hbar^2}} \int_0^{\infty} \frac{t(N_D, E)}{\exp[(E - E_{\text{FM}}(N_D))/k_B T] + 1} \, dE$$

where $m_e$, $q$, $\hbar$, $E$, $t(N_D, E)$, $N_D$, $E_{\text{FM}}(N_D)$, $k_B$, $T$ are the effective electron mass, electric charge, reduced Planck constant, energy, transmission coefficient, donor concentration, metal Fermi level, Boltzmann constant, and temperature, respectively. The donors are fully activated and $N_D = N_i$ in the neutral region (right-hand side of the energy band diagram shown in the inset of Figure 3) Assuming that the Schottky barrier height is 0.8 eV, we calculated $\rho_c$ at low temperature as a function of $N_c$ (plotted in Figure 3). The calculation results clearly suggest that contact resistivity is highly sensitive to the active carrier concentration in the contact layers. For example, increasing $N_c$ from $3 \times 10^{18}$ to $5 \times 10^{19} \, \text{cm}^{-3}$ improves $\rho_c$ by more than two orders of magnitude. Thus, maximization of active carrier concentration should be implemented to improve the contact resistivity. Note that the contact resistivity is also highly sensitive to the Schottky barrier height.

We attempted to maximize the active carrier concentration by optimizing the doping. As a reference structure, we fabricated an initial-side contact layer without $\delta$-doping. The initial-side contact layer consists of (1) 20 nm n-GaAs ($N_{\text{Si}}^{(1)} = 5 \times 10^{18} \, \text{cm}^{-3}$), (2) 15 nm n-GaAs ($N_{\text{Si}}^{(2)} = 3 \times 10^{18} \, \text{cm}^{-3}$), and (3) 10 nm GaAs (nondoped) grown at 530 °C. After the growth of the initial-side contact layer, an imitation device layer was grown on top at 580 °C. In this sample, the VDP measurement gave a total active carrier density ($N_{\text{total}}$) of 1.41 $\times 10^{13} \, \text{cm}^{-2}$ at RT. This density is very close to the total doped Si density of $1.45 \times 10^{15} \, \text{cm}^{-2}$, indicating that almost all the doped Si atoms had become active donors. Next, we checked $N_{\text{total}}$ in the initial-side contact layer with high-concentration Si ($N_{\text{Si}}^{(3)} = 5.5 \times 10^{19} \, \text{cm}^{-3}$) by adding $\delta$-doping ($N_{\text{DSi}} = 1 \times 10^{13} \, \text{cm}^{-2}$ and $M_\delta = 11$); $\delta$-doping was also performed on top of the sacrificial layer) with a 2 nm separation. Although we drastically increased the Si concentration, the active carrier density measured by the VDP method at RT rather decreased to $N_{\text{total}} = 1.33 \times 10^{13} \, \text{cm}^{-2}$. The reduction of active carrier density indicates that $N_{\text{Si}}^{(3)} = 5.5 \times 10^{19} \, \text{cm}^{-3}$ is too high, and the carrier compensation effects (formation of Si on As-site, Si–Si pairs, or Si–X centers) become dominant.23

To maximize the active carrier concentration, therefore, we investigated the lower Si-doping concentration conditions. We fabricated a series of samples in which initial-side contact layers consisted of (1) 20 nm n-GaAs ($N_{\text{Si}}^{(1)} = 5\sim17.5 \times 10^{18} \, \text{cm}^{-3}$), (2) 15 nm n-GaAs ($N_{\text{Si}}^{(2)} = 3 \times 10^{18} \, \text{cm}^{-3}$), and (3) 10 nm GaAs (nondoped). In the layer (1), $\delta$-doping of Si ($N_{\text{DSi}} = 0\sim5 \times 10^{12} \, \text{cm}^{-2}$) was performed with a 4 nm separation to realize a moderately high concentration of Si. The $\delta$-doping was also applied to the sacrificial layer ($M_\delta = 6$). After the growth, imitation device layers were grown on top at 580 °C. Figure 4 shows the carrier density of the samples at RT measured using van der Pauw (VDP) method. On increasing $N_{\text{Si}}^{(1)}$ from $0.5 \times 10^{19}$ to $1.25 \times 10^{19} \, \text{cm}^{-3}$, $N_{\text{total}}$ increases from $1.41 \times 10^{13}$ to $1.93 \times 10^{13} \, \text{cm}^{-2}$. However, further increasing $N_{\text{Si}}^{(1)}$ results in a decrease in $N_{\text{total}}$ to $1.67 \times 10^{13} \, \text{cm}^{-2}$. This result reveals that the active carrier density peaks when $N_{\text{Si}}^{(1)}$ is around $1.25 \times 10^{19} \, \text{cm}^{-3}$ ($N_{\text{DSi}} = 3 \times 10^{12} \, \text{cm}^{-2}$).
Using the sample with the maximum carrier concentration \((N_{Si}^{(1)} = 1.25 \times 10^{19} \text{ cm}^{-3})\), we evaluated \(\rho_c\) of the initial-side contact layer by the TLM measurement after the substrate inversion process, as shown in Figure 10c. From here, the thickness of the layer (1) is fixed at 28 nm, and \(\delta\)-doping of Si \((N_{Si}^{(1)} = 3 \times 10^{15} \text{ cm}^{-2})\) was performed with a 4 nm separation. No \(\delta\)-doping was performed on the sacrificial layers. The details of the structure of the initial-side contact layer are as follows: (1) 28 nm n-GaAs \((N_{Si}^{(1)} = 1.25 \times 10^{19} \text{ cm}^{-3})\), (2) 15 nm n-GaAs \((N_{Si}^{(2)} = 3 \times 10^{18} \text{ cm}^{-3})\), and (3) 10 nm GaAs (nondoped). As seen in Figure 5a, the \(I-V\) curve between the two electrodes exhibits a linear property at RT in a range of \(\pm 0.3\) V, indicating ohmic characteristics in this range. The value of \(\rho_c\) measured by TLM is \(2.0 \times 10^{-2} \text{ \Omega cm}^{2}\). At 77 K, however, \(\rho_c\) increases to 6.4 \(\text{ \Omega cm}^{2}\), and a nonlinear property was observed. The contact resistivity is therefore improved by optimizing the Si-doping concentration. However, it is still insufficient to realize an ohmic contact at low temperatures.

We attributed the observed difference in contact resistivity between the end-side and initial-side contacts to changes caused by annealing. Although the samples were cooled down immediately after the growth of the end-side contact layers, the initial-side contact layers are annealed at 580 °C during the growth of the imitation device layers. To suppress these temperature-induced changes, we also grew the upper imitation device layer at 530 °C for the next step. At 530 °C, we can still maintain the quality of GaAs and AlGaAs at a reasonably high level. By reducing the temperature for the imitation device layer growth from 580 °C to 530 °C, we found that \(N_{Si}^{total}\) measured by VDP slightly increased from 2.02 \(\times 10^{13}\) to 2.33 \(\times 10^{13} \text{ cm}^{-2}\). The increase in carrier density at a lower growth temperature is consistent with that reported in the literature. Figure 5b shows the \(I-V\) curves between the two electrodes at RT and 77 K of the sample in which the upper layers of the initial-side contact layers are annealed at 580 °C (red) and 530 °C (blue). The measurements were performed after the substrate inversion process. The inset shows a linear plot around the surfaces.

![Figure 5](image1)

**Figure 5.** \(I-V\) curves between two electrodes. The upper layers were grown at (a) 580 °C and (b) 530 °C. The size of the electrodes and separation between the two electrodes are 600 \(\times 400 \mu \text{m}^{2}\) and 100 \(\mu \text{m}\), respectively.

Of \(\pm 0.3\) V at 77 K are good enough for many GaAs-based plasmo-electronic device applications because their device core layers exhibit relatively high resistance.\(^{30}\) We additionally point out that further improvement of the contact properties might be obtained by tuning the Schottky barrier heights if necessary.\(^{31,32}\)

For the further understanding of the effects of the annealing temperature on contact resistivity, we checked the Si depth profiles using secondary ion mass spectrometry (SIMS). The SIMS measurements were performed on the samples after the substrate inversion process (Figure 10c) whose \(I-V\) curves are shown in Figure 5. Figure 6 shows the Si depth profiles. The high Si concentration observed at the surface region (0–3 nm) is due to surface contamination and is not discussed. On decreasing temperature from 580 °C to 530 °C, the lack of diffusion of Si into the nondoped GaAs layer is clearly visible. Suppression of diffusion causes an increase in Si concentration in the highly Si-doped layer (1) (+20%). Here, we additionally noted that Si diffusion into AlGaAs (the device layer) is highly suppressed in the sample grown at 530 °C. This is important for improving the device’s performance, such as by reducing the leakage current in the photodetectors.

Because the SIMS measurement cannot distinguish the activation state of Si atoms, we also evaluated the depth profiles of active carrier density by the VDP measurements...
combined with step-by-step etching (the details are explained in the Experimental Section below). In addition to the two samples of initial-side contacts (whose upper device imitation layers were grown at 530 °C and 580 °C) after the substrate inversion process, we also performed the same experiment on the end-side contact in which the same layer structure was grown at 530 °C (1) 10 nm GaAs (nondoped), (2) 15 nm n-GaAs \((N_{Si}^{(1)} = 3 \times 10^{18} \text{ cm}^{-3})\), and (3) 28 nm n-GaAs \((N_{Si}^{(1)} = 1.25 \times 10^{19} \text{ cm}^{-3} \text{ at } 530 ^\circ \text{C})\). Figure 7 shows the plot of the reduction of \(N_{total} (\Delta N_{total})\) by each 2 nm-thick etching step. It is clear that the active carrier density of the top part of the initial-side contact in which the upper layer was grown at 530 °C is higher than that of the initial-side contact in which the upper layer was grown at 580 °C (+60–80%). This carrier density difference is larger than the Si concentration difference (+20%), which is possibly due to the inactivation of doped Si (compensation effects) by high temperature annealing in addition to the effects of diffusion \(^{29,33}\). From the results, it is reasonable to conclude that the actual carrier volume density in the highly Si-doped layer (1) increased (+60–80%) by reducing the growth temperature. According to our calculations of the contact resistivity in the above section, the change is likely to result in a significant change in the contact resistivity.

Another noteworthy feature observed in Figure 7 is that the carrier concentration in the top part of the end-side contact is much higher than those in the initial-side contacts. Most probably, this high carrier concentration is the main reason why it is easy to obtain a nonalloyed ohmic contact to the end side. During the growth of Si-doped GaAs, surface segregation of Si occurs, resulting in the high Si concentration at the top part of the end-side contact. \(^{29,33}\) In addition, the end-side contact is cooled down immediately after the growth, which suppresses the inactivation of Si (compensation effects). \(^{29,33}\) In contrast, the surface segregation of Si reduces the Si concentration at the surfaces of initial-side contacts. Also, the initial-side contacts are inevitably annealed for the growth of upper layers. These make it difficult to obtain low-resistance Ohmic properties at the initial-side contacts. Therefore, decreasing the growth temperature for the contact layers and upper layers is highly important for minimizing these effects while the crystal quality of the upper layer should also be taken into account.

To minimize \(\rho_c\), we investigated \(\rho_c\) as a function of \(N_{Si}^{(1)}\) controlled by the \(\delta\)-doping density per layer \((N_{2DSi})\). Except for the Si concentration, the growth conditions are the same as with the sample shown in Figure 5b (the upper layers were grown at 530 °C). Figure 8a shows \(\rho_c\) of samples with various

![Figure 7](image1)

**Figure 7.** Plot of active carrier density difference \(\Delta N_{total}\) of the initial-side contacts in which the upper layers of the initial-side contact layers are grown at 580 °C (red), 530 °C (blue), and end-side contact (black). The measurements for the initial-side contacts were performed after the substrate inversion process.

![Figure 8](image2)

**Figure 8.** (a) Contact resistivity \(\rho_c\) at RT and 77 K and (b) total active carrier density \(N_{total}\) at RT as a function of the Si concentration in layer (1) \(N_{Si}^{(1)}\). When \(\rho_c\) is higher than \(-1.5 \times 10^{-1} \Omega \cdot \text{cm}^2\), nonlinear \(I–V\) curves were observed, indicating a Schottky-like contact.

\(N_{Si}^{(1)}\) at RT and 77 K. When \(N_{Si}^{(1)}\) is tuned in a range of 0.75 \(\times 10^{19}\) to 12.5 \(\times 10^{19} \text{ cm}^{-3}\), nonalloyed ohmic contacts are realized down to 77 K in a range of ±0.3 V. In contrast, with higher or lower \(N_{Si}^{(1)}\), \(\rho_c\) exhibits high values. Optimal contact resistivity was obtained at \(N_{Si}^{(1)} = 7.5 \times 10^{18} \text{ cm}^{-3}\) \((\rho_c = 8.5 \times 10^{-3} \Omega \cdot \text{cm}^2 \text{ at RT and } \rho_c = 4.1 \times 10^{-2} \Omega \cdot \text{cm}^2 \text{ at 77 K})\), where \(N_{2DSi}\) per layer is \(1 \times 10^{19} \text{ cm}^{-2}\). Figure 9 shows the \(I–V\) curves between the two electrodes at RT and 77 K of the optimal initial-side contact. Although small nonlinearity is still visible in

![Figure 9](image3)

**Figure 9.** \(I–V\) curves between two electrodes at RT and 77 K. \(N_{Si}^{(1)}\) is \(7.5 \times 10^{18} \text{ cm}^{-3}\) and the upper layers were grown at 530 °C. The size of the electrodes and separation between the two electrodes are 600 \(\times 400 \mu\text{m}^2\) and 100 \(\mu\text{m}\), respectively.
the high voltage region at 77 K, ohmic property is further improved compared with Figure 5b.

Finally, we would like to note an additional factor that has significant effects on contact resistivity. The total active carrier density of the sample with $N_{\text{Si}}^{(1)} = 7.5 \times 10^{18}$ cm$^{-3}$ (which exhibits the best contact resistivity) is $N_{\text{total}} = 2.09 \times 10^{13}$ cm$^{-2}$, which is lower than that of the sample with $N_{\text{Si}}^{(2)} = 1.25 \times 10^{19}$ cm$^{-3}$ ($N_{\text{total}} = 2.33 \times 10^{13}$ cm$^{-2}$), as shown in Figure 5b. In addition, the total active carrier density of the sample with $N_{\text{Si}}^{(2)} = 1.25 \times 10^{19}$ cm$^{-3}$ is higher than that of the sample with $N_{\text{Si}}^{(2)} = 5 \times 10^{18}$ cm$^{-3}$ ($N_{\text{total}} = 1.84 \times 10^{13}$ and $1.59 \times 10^{13}$ cm$^{-2}$). However, the contact resistivity at RT of the former sample is higher than that of the latter sample ($\rho_{c} = 2.2 \times 10^{-1}$ cm$^{-2}$) as compared with $5.7 \times 10^{-2}$ cm$^{-2}$. It has been reported that high doping of Si and/or annealing results in changes in the Si configurations in the GaAs.23,33,35,36 These changes possibly modify the electronic structures of n-GaAs, which in turn would affect the contact resistivity. To verify if this is the case, however, further studies will be needed.

Hence, tuning $N_{\text{Si}}^{(1)}$ to within a narrow range of $7.5 \times 10^{18}$ to $1.25 \times 10^{19}$ cm$^{-3}$ and maintaining a slightly low substrate temperature of 530 °C are the keys to the creation of an initial-side nonalloyed ohmic contact.

Although the obtained contact resistivities are not as low as the typical values ($\rho_{c} = 10^{-6}$ to $10^{-7}$ Ω·cm$^{-2}$) for the alloyed ohmic contacts to n-GaAs and small nonlinearity appears in the high voltage region at 77 K, the values are still in the acceptable range for specific applications such as metasurface QWIPs.

**CONCLUSIONS**

We have realized nonalloyed ohmic contacts on buried n-GaAs that appear after bonding the original substrates to support substrates with an upside-down configuration, followed by etching the substrates and sacrificial layers. An end-side nonalloyed ohmic contact is easily obtained by doping with ultrahigh density Si as reported previously. When similar structures are applied to the initial-side contact, however, the contacts become a Schottky-like contact. For the initial side, we found that moderate Si-doping density, which is lower than that for the end side, is necessary. The temperature for the growth of upper device layers on the initial-side contact layers is critically important for the suppression of Si diffusion and inactivation. Nonalloyed ohmic contacts in a range of ±0.3 V that operate from RT to 77 K can be obtained by tuning the concentration of doped Si to within a narrow range of $N_{\text{Si}} = 7.5 \times 10^{18} \pm 1.25 \times 10^{19}$ cm$^{-3}$ and reducing the substrate temperature during the growth of the upper device layers from 580 to 530 °C. The optimal contact resistivity obtained in this study is $\rho_{c} = 8.5 \times 10^{-3}$ Ω·cm$^{-2}$ at RT and $\rho_{c} = 4.1 \times 10^{-2}$ Ω·cm$^{-2}$ at 77 K for $N_{\text{Si}} = 7.5 \times 10^{18}$ cm$^{-3}$. Our results show that double-sided nonalloyed ohmic contacts to Si-doped GaAs are feasible in which ohmic contacts are realized on both sides without the formation of an intermediate alloyed layer for metal–semiconductor–metal plasmorelectronic device applications. The required thickness for a contact layer is only 53 nm, which contributes to the realization of thin semiconductor layers enabling higher field enhancement.

**EXPERIMENTAL SECTION**

The samples were grown on semi-insulating GaAs (100) substrates using a solid-source molecular beam epitaxy (MBE) system. Typical sample structures for the evaluation of initial-side contacts are shown schematically in Figure 10a. After the growth of GaAs buffer and Al$_{0.9}$Ga$_{0.1}$As (and Al$_{0.55}$Ga$_{0.45}$As) sacrificial layers, the n-GaAs initial-side contact layer was
grown at 530 °C. The contact layer consists of (1) a highly Si-doped GaAs layer (Si concentration \( N_Si \geq 5 \times 10^{18} \text{ cm}^{-3} \)), (2) n-GaAs (\( N_Si = 3 \times 10^{18} \text{ cm}^{-3} \)), and (3) GaAs (nondoped). To minimize Si segregation from the layer (1) to the device layer grown on top of them, we inserted layers (2) and (3). Next, a device layer (or imitation device layer) consisting of Al\(_x\)Ga\(_{1-x}\)As and GaAs layers was grown at 580 °C or 530 °C. Note that 580 °C is the typical growth temperature for high-quality GaAs and AlGaAs by MBE.37–39 To increase the number of Si atoms substituted for Ga, the layer (1) was grown at somewhat low temperature of 530 °C, and a \( \delta \)-doping technique was used to increase Si.39,40 During the growth of n-GaAs (\( N_Si = 5 \times 10^{18} \text{ cm}^{-3} \)), we interrupted the growth and performed \( \delta \)-doping of Si (\( N_{\text{DSSi}} = 0–1 \times 10^{13} \text{ cm}^{-2} \)) with 2–4 nm separations.

The total active carrier density (\( N_{\text{total}} \text{ cm}^{-2} \)) in the contact layers (layers (1), (2), and (3)) was checked using conventional van der Pauw (VDP) methods. For the measurements, InSn was thermally diffused from the top surfaces at 420–450 °C, as shown in Figure 10b.

For evaluation of contact resistivity, the original substrates were bonded to the support GaAs substrates using glue upside down (Figure 10c-1), followed by removal of the GaAs original substrates and AlGaAs sacrificial layers by mechanical and chemical etching (Figure 10c-2). Ti/Au metals were then evaporated to the exposed initial-side contact layers (1) for electrode formation (Figure 10c-3). The samples were not annealed after metal deposition. The \( I-V \) curves are obtained in between the two electrodes. The contact resistivity (\( \rho_{\text{c}} \Omega \text{ cm}^{2} \)) was estimated using the transmission line model (TLM) measurement.41 The values of \( \rho_{\text{c}} \) were estimated by fitting the values of resistance between the electrodes with different separations.

The depth profiles of the doped Si density were measured using secondary ion mass spectrometry (SIMS). The depth profiles of the active carrier density were evaluated by the VDP method combined with step-by-step etching of the sample surfaces by 2.3% tetramethylammonium hydroxide solution. Each time after the 2 nm-thick etching (the etching time was 1 min), we measured the total active carrier density (\( N_{\text{total}} \text{ cm}^{-2} \)) by the VDP measurement. By calculating the difference of \( N_{\text{total}} \) before and after each etching step (\( \Delta N_{\text{total}} \)), we evaluated the active carrier area density in the etched 2 nm-thick layers.

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## REFERENCES

1. Koenderink, A. F.; Alù, A.; Polman, A. Nanophotonics: Shrinking light-based technology. Science 2015, 348, 516–521.

2. Gramotnev, D. K.; Bozhevolnyi, S. I. Plasmonics beyond the diffraction limit. Nat. Photonics 2010, 4, 83–91.

3. Miyazaki, H. T.; Kurokawa, Y. Squeezing Visible Light Waves into a 3-nm-Thick and 55-nm-Long Plasmon Cavity. Phys. Rev. Lett. 2006, 96, No. 097401.

4. Vedantam, S.; Lee, H.; Tang, J.; Conway, J.; Staffaroni, M.; Yablonovitch, E. A Plasmonic Dimple Lens for Nanoscale Focusing of Light. Nano Lett. 2009, 9, 3447–3452.

5. Im, H.; Bantz, K. C.; Lindquist, N. C.; Haynes, C. L.; Oh, S. H. Vertically Oriented Sub-10-nm Plasmonic Nanopad Arrays. Nano Lett. 2010, 10, 2231–2236.

6. Miyazaki, H. T.; Kasaya, T.; Iwanaeg, M.; Choi, B.; Sugimoto, Y.; Sakoda, K. Dual-band infrared metasurface thermal emitter for CO\(_2\) sensing. Appl. Phys. Lett. 2014, 105, No. 121107.

7. Russell, K. J.; Liu, T. L.; Cui, S.; Hu, E. L. Large spontaneous emission enhancement in plasmonic nanocavities. Nat. Photonics 2012, 6, 459–462.

8. Williams, B. S.; Kumar, S.; Callebaut, H.; Hu, Q.; Reno, J. L. Terahertz quantum-cascade laser at \( 2\times10 \mu \text{m} \) using metal waveguide for mode confinement. Appl. Phys. Lett. 2003, 83, 2124–2126.

9. Fan, J. A.; Belkin, M. A.; Capasso, F.; Khanna, S. P.; Lachab, M.; Davies, A. G.; Linfield, E. H. Wide-ridge metal-metal terahertz quantum cascade lasers with high-order lateral mode suppression. Appl. Phys. Lett. 2008, 92, No. 031106.

10. Fatholouloumi, S.; Dupont, E.; Razaviopour, S. G.; Loframboise, S. R.; Parent, G.; Wasilewski, Z.; Liu, H. C.; Ban, D. On metal contacts of terahertz quantum cascade lasers with a metal–metal waveguide. Semicond. Sci. Technol. 2011, 26, No. 105021.

11. Li, Q.; Li, Z.; Li, N.; Chen, X.; Chen, P.; Shen, X.; Lu, W. High-polarization-discriminating infrared detection using a single quantum well sandwiched in plasmonic micro-cavity. Sci. Rep. 2014, 4, No. 6332.

12. Palaferrri, D.; Todorov, Y.; Bigioli, A.; Mottaghizadeh, A.; Gacemi, D.; Calabrese, A.; Vasaneli, A.; Li, L.; Davies, A. G.; Linfield, E. H.; Kapalsidis, F.; Beck, M.; Faist, J.; Sirtori, C. Room-temperature nine-\( \mu \text{m} \)-wavelength photodetectors and GHz-frequency heterodyne receivers. Nature 2018, 556, 85–88.

13. Palaferrri, D.; Todorov, Y.; Chen, Y. N.; Madeo, J.; Vasaneli, A.; Li, H.; Davies, A. G.; Linfield, E. H.; Sirtori, C. Patch antenna terahertz photodetectors. Appl. Phys. Lett. 2015, 106, No. 161102.

14. Miller, D. C. The alloying of gold and gold alloy ohmic contact Metalizations with gallium arsenide. J. Electrochem. Soc. 1980, 127, 467–475.

15. Braslau, N. Alloyed ohmic contacts to GaAs. J. Vac. Sci. Technol. 1981, 19, 803–807.

16. Ogawa, M. Alloving behavior of Ni/Au-Ge films on GaAs. J. Appl. Phys. 1980, 51, 406–412.

17. Momme, D.; Pan, E. T.−S.; Murakami, M. Uniform and thermally stable AuGeNi ohmic contacts to GaAs. Appl. Phys. Lett. 1985, 46, 1141–1143.

18. Shen, T. C.; Gao, G. B.; Morkoç, H. Recent developments in ohmic contacts for III–V compound semiconductors. J. Vac. Sci. Technol., B 1992, 10, 2113–2132.

19. Zheng, L. R.; Wilson, S. A.; Lawrence, D. J.; Rudolph, S. I.; Chen, S.; Braunstein, G. Shallow ohmic contacts ton—type GaAs and Al\(_x\)Ga\(_{1−x}\)As. Appl. Phys. Lett. 1992, 60, 877–879.
(20) Zheng, L. Shallow ohmic contact formation by sequential deposition of Pd/AuGe/Au on GaAs and rapid thermal annealing. J. Appl. Phys. 1992, 71, 3566−3571.

(21) Chai, Y. G.; Chow, R.; Wood, C. E. C. The effect of growth conditions on Si incorporation in molecular beam epitaxial GaAs. Appl. Phys. Lett. 1981, 39, 800−803.

(22) Akimoto, K.; Dohsen, M.; Arai, M.; Watanabe, N. As/Ga flux ratio dependence on Si incorporation in molecular beam epitaxial GaAs. Appl. Phys. Lett. 1983, 43, 1062−1064.

(23) Maguire, J.; Murray, R.; Newman, R. C.; Beall, R. B.; Harris, J. J. Mechanism of compensation in heavily silicon-doped gallium arsenide grown by molecular beam epitaxy. Appl. Phys. Lett. 1987, 50, 516−518.

(24) Kirchner, P. D.; Jackson, T. N.; Pettit, G. D.; Woodall, J. M. Low-resistance nonalloyed ohmic contacts to Si-doped molecular beam epitaxial GaAs. Appl. Phys. Lett. 1985, 47, 26−28.

(25) Schubert, E. F.; Cunningham, J. E.; Tsang, W. T.; Chiu, T. H. Delta-doped ohmic contacts to n-GaAs. Appl. Phys. Lett. 1986, 49, 292−294.

(26) Patkar, M. P.; Chin, T. P.; Woodall, J. M.; Lundstrom, M. S.; Melloch, M. R. Very low resistance nonalloyed ohmic contacts using low-temperature molecular beam epitaxy of GaAs. Appl. Phys. Lett. 1995, 66, 1412−1414.

(27) Kaminska, M.; Liliental-Weber, Z.; Weber, E. R.; George, T.; Kortright, J. B.; Smith, F. W.; Tsaur, B-Y.; Calawa, A. R. Structural properties of As-rich GaAs grown by molecular beam epitaxy at low temperatures. Appl. Phys. Lett. 1989, 54, 1881−1883.

(28) Cho, S. M.; Lee, J. D.; Lee, H. H. Specific resistivity of ohmic contacts to n-type direct band-gap III-V compound semiconductors. J. Appl. Phys. 1991, 70, 282−287.

(29) Silveira, J. P.; Briones, F. Low temperature growth of highly doped GaAs: Si by atomic layer molecular beam epitaxy. Appl. Phys. Lett. 1994, 65, 573−574.

(30) Chen, Y. N.; Todorov, Y.; Askenazi, B.; Vasanelli, A.; Biasiol, G.; Colombelli, R.; Sirtori, C. Antenna-coupled microcavities for enhanced infrared photo-detection. Appl. Phys. Lett. 2014, 104, No. 031113.

(31) Hu, J.; Saraswat, K. C.; Wong, H.-S. P. Metal/III-V Schottky barrier height tuning for the design of nonalloyed III-V field-effect transistor source/drain contacts. J. Appl. Phys. 2010, 107, No. 063712.

(32) Kim, S.-H.; Kim, G.-S.; Kim, S.-W.; Kim, J.-K.; Choi, C.; Park, J.-H.; Choi, R.; Yu, H.-Y. Non-alloyed ohmic contacts on GaAs using metal-interlayer-semiconductor structure with SF6 plasma treatment. IEEE Electron Device Lett. 2016, 37, 373−376.

(33) Ishikawa, T.; Inata, T.; Kondo, K.; Shibatomi, A. Annealing effect in Si-doped GaAs and AlGaAs layers grown by MBE. Electron. Lett. 1986, 22, 189−190.

(34) Andrieu, S.; d’Avitaya, F. A.; Pfister, J. C. Surface segregation mechanism during two-dimensional epitaxial growth: The case of dopants in Si and GaAs molecular-beam epitaxy. J. Appl. Phys. 1989, 65, 2681−2687.

(35) Williams, E. W. Evidence for self-activated luminescence in GaAs: The gallium vacancy-donor center. Phys. Rev. 1968, 168, 922−928.

(36) Newman, R. C. The lattice locations of silicon impurities in GaAs: effects due to stoichiometry, the Fermi energy, the solubility limit and DX behaviour. Semicond. Sci. Technol. 1994, 9, 1749−1762.

(37) Hiyamizu, S.; Fuji, T.; Mimura, T.; Nanbu, K.; Saito, J.; Hashimoto, H. The effect of growth temperature on the mobility of two-dimensional electron gas in selectively doped GaAs/N-AlGaAs heterostructures grown by MBE. Jpn. J. Appl. Phys. 1981, 20, L455−L458.

(38) Morkoç, H.; Drummond, T. J.; Kopp, W.; Fischer, R. Influence of substrate temperature on the morphology of AlGa1−xAs grown by molecular beam epitaxy. J. Electrochem. Soc. 1982, 129, 824−826.

(39) Leonard, D.; Krishnamurthy, M.; Reeves, C. M.; Denbaars, S. P.; Petroff, P. M. Direct formation of quantum-sized dots from uniform coherent islands of InGaAs on GaAs surfaces. Appl. Phys. Lett. 1993, 63, 3203−3205.