Increased dephasing length in heavily doped GaAs

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Abstract

Ion implantation of S and Te followed by sub-second flash lamp annealing with peak temperature about 1100 °C is employed to obtain metallic \(n^{++}\)-GaAs layers. The electron concentration in annealed GaAs is as high as \(5 \times 10^{19} \text{cm}^{-3}\), which is several times higher than the doping level achievable by alternative methods. We found that heavily doped \(n^{++}\)-GaAs exhibits positive magnetoconductance in the temperature range of 3–80 K, which is attributed to the magnetic field suppressed weak localization. By fitting the magnetoconductance results with the Hikami–Larkin–Nagaoka model, it is found that the phase coherence length increases with increasing carrier concentration at low temperature and is as large as 540 nm at 3 K. The temperature dependence of the phase coherence length follows \(l_\eta \propto T^\eta (\eta \sim 0.3)\), indicating defect-related scattering as the dominant dephasing mechanism. In addition, the high doping level in \(n\)-type GaAs provides the possibility to use GaAs as a plasmonic material for chemical sensors operating in the infrared range.

1. Introduction

In disordered electron systems, the electrons can be localized, which leads to weak localization (WL) and eventually Anderson insulators [1]. WL is a quantum mechanical effect where the constructive interference occurs between two electron waves traveling along a closed path in an opposite direction, leading to a suppression of conductivity. Under the action of a magnetic field, the two waves acquire a phase difference and the interference conditions are violated and causing a positive magnetoconductance (PMC, or negative magnetoresistance (NMR)). The theory of PMC in doped semiconductor was proposed decades ago [2] and explained by WL on the metallic side of the insulator–metal transition (IMT) [3, 4]. PMC has already been reported in \(\delta\)-doped layers in GaAs [5], GaAs quantum wells [6] and GaAs/Al\(_{x}\)Ga\(_{1-x}\)As heterostructures [7]. In the above-mentioned cases, the large amount of carriers are confined within a thin layer. However, unlike heavily doped Si and Ge epilayers [4, 8], PMC in GaAs epilayers has not yet been reported in metallic-like \(n\)-type GaAs with carrier concentration above \(10^{19} \text{cm}^{-2}\), due to the \(n\)-type doping limitation. Moreover, although PMC effect has been reported in \(n\)-type GaAs film in the IMT regime with carrier concentration in the order of \(10^{18} \text{cm}^{-3}\), it is highly dependent on electron–electron interaction effect including orbital and Zeeman contributions [3, 9]. Specifically, Monsterle et al, have showed a PMC effect in \(n\)-GaAs close to the metal–insulator transition, in which the electron concentration is \(2.9 \times 10^{19} \text{cm}^{-3}\) [3]. In addition, Gilbertson et al, have reported a WL with a 2D to 3D crossover in \(n\)-type GaAs with the doping level of \(4 \times 10^{17} \text{cm}^{-2}\) as temperature increases from 2 to 50 K [9]. The reason for the doping limitation in GaAs is mainly due to the compensating effect. Si is the most common \(n\)-type dopant for GaAs. Unfortunately, Si in GaAs is an amphoteric dopant, which limits the effective electron concentration to the
level of about $5 \times 10^{18} \text{cm}^{-3}$ [10, 11]. Specifically, for high doping level, Si occupies both Ga and As sites, donating both electrons and holes simultaneously, which causes compensation effect and decreases the effective carrier concentration.

A reasonable alternative for $n$-type doping in GaAs is choosing shallow donors from group VI, i.e. chalcogens like S, Se or Te [12–15]. Chalcogens in GaAs are located at the substitutional positions replacing As and donate electrons. But it is rather surprising that the limit of the electron concentration of around $10^{19} \text{cm}^{-3}$ is a universal feature existing in group VI donors as well, although it is higher than in Si-doped GaAs. The saturation of the electron concentration for chalcogen doped GaAs is due to intrinsic defects, mainly triple negatively charged gallium vacancies ($V_{Ga}^-$) bonded with donors [16]. With increasing doping level, the Fermi energy shifts up toward the conduction band, reducing the formation energy of $V_{Ga}$. This effect increases the number of $V_{Ga}$, which compensates the donors and leads to a saturation of the electron concentration [12].

In this work, ion implantation followed by sub-second annealing, in particular millisecond (ms)-range flash lamp annealing (FLA) is used to achieve high doping levels in GaAs. The ultra-doped GaAs layers are prepared by implanting S and Te (as donors), and Zn (as acceptors) ions into intrinsic GaAs wafers. After ion implantation, samples are annealed by FLA for 20 ms with a peak temperature of around 1100 °C. $n^{++}$ doping for GaAs in the range of $10^{19} \text{cm}^{-3}$ has been realized for both S and Te dopants. The highest obtained electron concentration for S doped GaAs reaches $5 \times 10^{19} \text{cm}^{-3}$, which is much above the equilibrium solid solubility of S in GaAs ($\sim10^{19} \text{cm}^{-3}$) [12], and it is three orders of magnitude higher than the room-temperature critical carrier concentration of $1.6 \times 10^{19} \text{cm}^{-3}$ required for the IMT in GaAs [17]. The $n^{++}$-GaAs (100) layer of around 100 nm thickness exhibits PMC. The magnetoconductance can be well fitted with the Hikami–Larkin–Nagaoka (HLN) model, which allows for the estimation of the phase coherence length ($l_\phi$). At 3 K $l_\phi$ is estimated as large as 540 nm for both S and Te doped GaAs, which is substantially longer than the previously reported values in other semiconductors, e.g. $\sim325 \text{nm}$ in Ge at 1.6 K and $\sim100 \text{nm}$ in ZnO at 5 K [4, 18, 19]. In addition, this unprecedented high electron concentration expectedly meets the requirement for GaAs as a plasmonic material with lower material losses due to a smaller effective mass compared with other heavily doped semiconductors [20].

2. Experimental details

Semi-insulating (100) GaAs wafers were implanted at room temperature with S, Te and Zn ions with energies of 55, 160 and 100 keV, respectively. The wafers were tilted by 7° with respect to the ion beam to avoid channeling effect. Before ion implantation, the required ion fluences and the implantation energies, which determine the dopant concentration and distribution in the implanted layer, were calculated using stopping range of ions in matter (SRIM) simulations [21]. Figure 1 shows the concentration profiles of Zn, S and Te with the highest fluence in GaAs obtained by SRIM-code simulation. We assumed that the heavily doped layer has the average thickness of about 100 ± 10 nm, which corresponds to the ion distribution where the dopant concentration decreases to 10% of the maximum concentration (marked with gray area). The peak concentrations of the implanted dopants are in the range of $3 \times 10^{19}–2 \times 10^{20} \text{cm}^{-3}$. The detailed information about the sample parameters is listed in table 1. After ion implantation but prior to FLA, a 100 nm thick SiO$_2$ layer was deposited by plasma enhanced chemical vapor deposition at 200 °C. The SiO$_2$ capping layer was used to prevent the decomposition of GaAs during the high temperature annealing process. After annealing, the capping layer was removed by wet etching in 10% HF:H$_2$O solution. GaAs samples were annealed by FLA at an energy density of 115 J·cm$^{-2}$ for 20 ms, which corresponds to a peak temperature of about 1100 °C. The details about the FLA system used to anneal implanted samples and the calculation of the temperature during FLA process can be found in reference [22]. Note that millisecond-range annealing can suppress the dopant diffusion in GaAs, which is reported in our previous work [23]. It means that the thickness of doped layer after FLA remains at around 100 nm. The energy difference between the Fermi level and the conduction band ($n$-type) or the valence band ($p$-type) resulting from the band filling in degenerated GaAs is deduced from equation (6).

The Rutherford backscattering spectrometry (RBS) measurements were performed with a collimated 1.7 MeV He$^+$ beam of the Rossendorf van de Graaff accelerator with a 10–20 nA beam current at a backscattering angle of 170°. The crystalline quality of heavily doped GaAs layers before and after FLA were estimated from random and channeling RBS spectra. The channeling spectra were collected by aligning the sample to make the impinging He$^+$ beam parallel to the GaAs [001] axes. The structural properties of implanted and annealed samples were also investigated by micro-Raman spectroscopy. The Raman spectra were collected in a backscattering geometry in the range of 100 to 600 cm$^{-1}$ with the resolution of 0.1 cm$^{-1}$ using a 532 nm Nd:YAG laser and a liquid nitrogen cooled silicon based charge-coupled device camera.
The laser power was 3.2 mW and the diameter of the laser spot was 0.8 μm. The spectra are taken in Z(YY)Z configuration. The optical properties were investigated by PL under a continuous wave green laser (532 nm) excitation with a maximum excitation power up to 10 mW with the diameter of the laser spot of 0.8 μm. The PL signal was collected with a liquid nitrogen cooled InGaAs detector and optical spectrometer with the resolution of 0.2 nm. The focal length of the spectrometers for both Raman and PL is 800 mm. Electrical properties were measured in van der Pauw geometry using a commercial Lakeshore Hall system with a magnetic field perpendicular to the sample plane and swept from −4 T to 4 T in the temperature range of 3–300 K. The electrical contacts for Hall effect measurements were made by indium. According to current–voltage curves, all contacts were confirmed to be ohmic.

3. Results and discussion

3.1. Structural properties

Figure 2 shows the RBS random and channeling (RBS/R and RBS/C) spectra for Zn, S and Te implanted GaAs samples with the highest doses before and after annealing. A non-implanted virgin GaAs is included as the reference. Since RBS/R spectra for all the samples including virgin GaAs are overlapping, only one representative spectrum is shown. The RBS/C spectra for as-implanted samples exhibit a broad damage peak with higher backscattering yields at the sample surface (from 1260 to 1370 keV), which is attributed to lattice damage introduced by the ion implantation. After FLA, the RBS/C of Te-doped sample reveals a minimum backscattering yield \( \chi_{\text{min}} \) (the ratio of the aligned to random yields) of about 11.0%, which is comparable to the value of 7.8% determined for the virgin GaAs samples. The slightly higher yield \( \chi_{\text{min}} \) can be due to a bit lower crystal quality than the virgin one. The RBS/C spectra obtained from samples doped with S and Zn after annealing overlap with the spectrum of the virgin GaAs in the near surface region, indicating a complete recrystallization of the implanted layer. For virgin GaAs and implanted samples after
annealing, in all RBS/C spectra the Ga (1320 keV) and As (1350 keV) related surface peaks are well-separated, which confirms the formation of high quality layers. The presented results show that the non-equilibrium millisecond-range FLA can effectively recrystallize the ion implanted GaAs without surface decomposition which is otherwise commonly observed after conventional high temperature annealing (above 800 °C) [24].

3.2. Optical properties

Figures 3(a) and (b) show the normalized Raman spectra from virgin, as-implanted and annealed GaAs samples. The Raman spectra of as-implanted samples (cyan curves) exhibit a broad phonon band, which indicates that the doped layers are totally amorphized. This is in good agreement with RBS results. After FLA, the transverse optical (TO) and longitudinal optical (LO) phonon modes clearly appear for all samples. Due to the selection rules of the Raman scattering, only the LO phonon mode should be detected in (100) GaAs [25, 26], as shown for virgin GaAs in the lower panel of figure 3(a). However, for annealed samples, the observed phonon mode located at the position of the TO phonon (about 268 cm$^{-1}$) is more prominent than the peak at the LO phonon mode position. This is due to the coupling between the LO phonon mode and the carrier-related plasmon mode, i.e. the coupled LO-phonon plasmon mode (CLOPM) [27–29]. In polar semiconductors like GaAs, the LO phonon couples strongly with the collective oscillations of the free-carrier system (plasmons). Figure 3(a) shows the experimental and fitting results using Gaussian deconvolution for Zn-doped GaAs samples. The broadening and the increase of the intensity of the CLOPM mode were observed with increasing doping/carrier concentration. The Stokes Raman scattering rate by the CLOPM mode can be written in the long-wavelength limit by taking into account both the allowed deformation potential and electro-optical contribution as follows [27, 30]:

\[
I_s(\omega) = A \left( n_\omega + 1 \right) \frac{\left[ \omega_{\text{TO}}^2 \left( 1 + C \right) - \omega^2 \right]^2}{\left( \omega_{\text{TO}}^2 - \omega^2 \right)} \times \text{Im} \left[ \frac{-1}{\varepsilon(\omega)} \right],
\]

where $A$ is an $\omega$-independent constant factor, $n_\omega$ is the Bose–Einstein distribution, $C$ is the Faust–Henry coefficient, $\omega$ is the angular frequency, $\omega_{\text{TO}}$ is the frequency of the TO mode, and $\varepsilon(\omega)$ is the dielectric function, which is given by [30]:

\[
\varepsilon(\omega) = \varepsilon_{\infty} \left[ 1 + \frac{\omega_{\text{LO}}^2 - \omega_{\text{TO}}^2}{\omega_{\text{TO}}^2 - \omega^2 - i\gamma\omega} \right] \frac{\omega_{\text{P}}^2}{\omega^2 + i\Gamma P \omega},
\]

where $\varepsilon_{\infty}$ is the high-frequency dielectric constant, $\omega_{\text{LO}}$ is the frequency of the LO mode, $\gamma$ is the LO damping factor, and $\Gamma P$ is the plasma damping constant. $\omega_{\text{P}}$ is the plasma (screened) frequency, which can be obtained by:

\[
\omega_{\text{P}}^2 = \frac{ne^2}{m^*_e \varepsilon_{\infty} \varepsilon_0},
\]
Figure 3. (a) and (b) Show the Raman spectra obtained from virgin, as-implanted and annealed GaAs samples. (c) Shows the Raman shift of the phonon mode and the plasma frequency $\omega_p$ versus concentration for heavily doped $p$-type and $n$-type GaAs calculated using equations (3) and (4): green circles—upper branch of the CLOPM phonon mode; blue squares—lower branch of the CLOPM phonon mode; black diamonds—plasma frequency for $p$-type GaAs; red triangles—plasma frequency for $n$-type GaAs.

where $\varepsilon_0$ is the vacuum permittivity, $m_e^*$ is the average electron effective mass, and $n$ is the electron concentration. For $p$-type samples, $n$ is replaced by $p$ (hole concentration) and $m_e^*$ is replaced by the heavy hole effective mass $m_{hh}^*$.

According to the equations (1)–(3), $I_s(\omega) \propto p$, which explains the increase of the intensity of the CLOPM with increasing dopant concentration. Moreover, the weak TO phonon mode (blue curves)
observed in Zn-doped samples is due to the disorder effect originating from the high dose implantation. It causes the breakdown of the Raman selection rules [25, 26].

In the n-type GaAs case, see figure 3(b), the lower branch of the CLOPM mode is located exactly at the TO phonon mode position and is insensitive to the change of the carrier concentration. n-type III–V semiconductors are characterized by high carrier mobilities and low carrier effective masses. In this case, the lower branch ($\omega_{L-}$) and upper branch ($\omega_{L+}$) of CLOPM should be observed. A theoretical model applying for $\omega_{L\pm}$ can be described by equation (4) [31]

$$\omega_{L\pm}^2 = \frac{1}{2} \left( \omega_{LO}^2 + \omega_p^2 + \Gamma_p \gamma \right) \pm \frac{1}{2} \sqrt{\left( \omega_{LO}^2 + \omega_p^2 + \Gamma_p \gamma \right)^2 - 4\omega_p^2 \omega_{LO}^2},$$

where the damping constant $\Gamma_p$ of the plasma oscillation can be written as follows [24]:

$$\Gamma_p = \frac{e}{\mu m^* \tau},$$

where $\mu$ is the mobility of the free carriers, and $\tau$ is the momentum relaxation time. We assume that the phonon mode observed at the position of the TO phonon mode originates from $\omega_{L-}$, since $\omega_{L-}$ reaches the TO phonon energy at high carrier concentration. The upper-frequency mode ($\omega_{L+}$) evolves from a phonon-like to a plasmon-like character with increasing carrier concentration [31, 32], see figure 3(c). The calculated plasma frequencies, using equation (3), for $p$-type and $n$-type GaAs are also shown in figure 3(c). For Zn-doped GaAs, the CLOPM peak is much broader than for $n$-type GaAs, due to the higher plasmon damping $\Gamma_p$ ($\sim$700 cm$^{-1}$) in $p$-type GaAs. Therefore, in $p$-type GaAs only one overdamped CLOPM mode is found [29]. Due to the lower effective mass for electrons than for holes, with an identical carrier concentration, the plasma frequency ($\omega_p$) for $n$-type GaAs is higher than that for $p$-type GaAs. Hence, the tuning of the plasma frequency is easier in the $n$-type GaAs. This makes $n$-type GaAs a potential material for mid-infrared plasmonic chemical sensors [20, 33].

Figure 4 shows the normalized room-temperature PL spectra obtained from GaAs samples doped with 0.32% S, 0.42% Te or 0.34% Zn after green laser excitation. The PL spectrum of virgin GaAs is shown for reference. The penetration depth of the green laser in GaAs is deeper than the thickness of the doped layer, therefore the presented PL spectra are a superposition of the PL emission collected from the top implanted layer and from the undoped substrate. The PL emission from the undoped substrate is expected to be located at the same position as that observed from virgin GaAs. After deconvolution of the PL spectra with two Gaussian peaks, the PL emission from the doped and undoped layers can be distinguished, as visualized with shadow areas and green dash-dot lines (see figure 4). Note that the peak at about 900 nm from the $n$-type GaAs layer is attributed to defect-related emission [34]. It is clearly visible that the peak positions of the 0.32% S- and 0.42% Te-doped samples exhibit a blue shift with respect to the virgin GaAs (see the violet area in figure 4). On the contrary, the 0.34% Zn doped sample shows a red shift (see the red area in figure 4). The blue shift in $n$-type GaAs is mainly due to the filling of the conduction band with carriers leading to the Fermi level upshift above the conduction band minimum. Likewise, in $p$-type GaAs, the Fermi level downshifts below the valence band maximum. The energy difference between the Fermi level and the conduction band (or valence band) follows the equation (6) [35]

$$E_F - E_C(E_V) = \frac{\hbar^2}{2m^*_e}(3\pi^2 n)^{2/3},$$

where $n$ is the free carrier density and $m^*_e$ is the electron effective mass. For $p$-type samples, $n$ is replaced by $p$ (hole concentration) and $m^*_e$ is replaced by the heavy hole effective mass $m^*_h$. The calculation results are listed in table 1. Taking into account the position of the Fermi level, all the doped samples are degenerate semiconductors expecting to show metallic-like behavior.

High-level doping can also induce bandgap narrowing, which lowers the electron energies as compared to a non-interacting carrier system. The spatially redistributed electrons reduce the long-range Coulombic interaction energy. As a result, the interaction energy reduces the total energy of the electron system. The change in energy gap follows a $1/3$ power law of the doping concentration, $\Delta E_g \propto n^{1/3}$ [36]. The bandgap narrowing causes the downshift of the conduction band minimum. However, the band filling effect in $n$-type GaAs prevails causing the blue shift of the PL peak position [37, 38]. The red shift of the PL emission in heavily doped $p$-type GaAs (the radiative recombination of heavy holes and free electrons) is dominantly attributed to the bandgap narrowing effect, leading to the decrease of emitted photons energy [35, 36, 39–42].
Figure 4. Normalized room-temperature PL spectra obtained from 0.32% S, 0.42% Te or 0.34% Zn samples. The PL spectrum of virgin GaAs is shown for comparison.

Figure 5. (a) The carrier concentration and (b) the Hall mobility at 300 K as a function of dopant concentration obtained from implanted GaAs followed by FLA. (c) Temperature-dependence of the sheet resistance of doped GaAs in the temperature range of 3 K to 300 K.

3.3. Transport properties
Figure 5(a) shows the carrier concentration at 300 K as a function of dopant concentration obtained from Zn, S and Te implanted GaAs followed by FLA. In the case of p-type GaAs (Zn doped), the highest hole concentration is $10.9 \times 10^{19} \text{ cm}^{-3}$, while the maximum electron concentration for n-type GaAs doped with S is as high as $5.2 \times 10^{19} \text{ cm}^{-3}$. High p-type doping of GaAs is easier than n-type doping. On one hand, this is due to the higher solid solubility of acceptors than donors in GaAs. On the other hand, the higher formation energy for VAs in p-type GaAs than VGa in n-type GaAs decreases the probability for the formation of acceptor-vacancy (e.g. Zn-VAs) complexes compared with the formation of donor-vacancy complexes (e.g. S-VGa) [12, 43]. This is in agreement with previous reports that high hole concentrations can be relatively easily achieved in GaAs. For example, using Be as the acceptor in GaAs, hole concentration as high as $10^{21} \text{ cm}^{-3}$ is possible [44]. Therefore, the much higher hole concentration obtained from Zn-doped GaAs compared to the electron concentration measured from S or Te doped GaAs is not surprising.

The effective Bohr radius of shallow-donor bound electrons in n-type GaAs with effective mass of 0.067 $m_e$ is $a_B = 103 \text{ Å}$ and the critical concentration of donors for IMT in GaAs at room temperature is $n_c = 1.6 \times 10^{16} \text{ cm}^{-3}$ [17]. That means that all the investigated n-type GaAs samples are highly degenerate semiconductors and supposed to exhibit metallic behavior. Since the larger effective mass leads to smaller Bohr radii ($m^* \propto \frac{1}{m}$), for the Zn-doped p-type GaAs, the effective Bohr radius is about 15.6 Å. The critical
hole concentration for IMT at room temperature in p-type GaAs is around $4.1 \times 10^{18}$ cm$^{-3}$, based on the Mott theory of the IMT as described by equation (7) [36, 45]

$$n_c^{1/3} a_B = 0.26 \pm 0.05,$$

where $a_B$ is effective Bohr radius, and $n_c$ is the critical electron concentration. Therefore, the presented p-type GaAs samples are expected to be also degenerate semiconductors like n-type GaAs.

Figure 5(b) shows the carrier mobility at 300 K as a function of dopant concentration. The mobility for n-type GaAs and p-type GaAs is in the range of 283–733 and 42–28 cm$^2$ (V$^{-1}$s$^{-1}$), respectively. The decrease of the carrier mobility with increasing carrier concentration is mainly attributed to electron–electron scattering and carrier scattering by the impurities, which is typical for heavily doped semiconductors [46]. Figure 5(c) shows the temperature-dependence of the sheet resistance obtained from implanted GaAs followed by FLA. p-type GaAs with the lowest Zn concentration (0.07% Zn) exhibits a different behavior from other samples. This is due to the lower hole concentration. The other samples show nearly temperature-independent sheet resistivity as expected for degenerate semiconductors.

Figure 6 shows the magnetoconductance of n-type and p-type GaAs samples. The conductance data is described by equation (8) and shows the change in the conductivity of measured sample under magnetic field $(G(B))$ relative to the zero field $(G(0))$ conductivity

$$\Delta G = G(B) - G(0) = \frac{1}{\rho(B)} - \frac{1}{\rho(0)},$$

where $\rho(0)$ and $\rho(B)$ are the resistivity of measured samples without and with magnetic field. Obviously, at low temperature, the n-type GaAs samples show PMC at low magnetic field (0.3 T), while p-type GaAs exhibits negative MC in the whole magnetic field range. The theory of PMC is extensively studied for heavily doped semiconductors [8, 47]. In weakly disordered semiconductors, it is explained that the multiple random elastic scattering of electrons from impurities leads to the quantum interference phenomena [1, 18]. Quantum interference can introduce the WL effect, which significantly influences the temperature dependent electron transport mechanisms. WL arises because of the self-interference of time-reversed electronic waves backscattered from impurities. It leads to an enhanced probability of carrier backscattering, and therefore results in an enhanced resistivity. The magnetic field applied normal to the plane of carrier motion breaks the time reversal symmetry, and suppresses the WL, leading to a PMC or named negative magnetoresistance. Note that in n-type GaAs, the WL corrections were first developed in the limit of weak disorder, when $k_f l \gg 1$ ($k_f$ is the Fermi wave length, and $l$ is the mean free-path). For our n-type GaAs shown in this manuscript, $k_f l = 11–26$ (estimated from the free-electron theory, $k_f l = h(3\pi^2)^{2/3}e^2/\rho m^*\hbar$) [48].

There are alternative two- and three-dimensional fitting models for n-type GaAs. The 3D model can be written as [1]:

$$\frac{\Delta \rho}{\rho} = -\alpha e^2 \frac{c^2}{2\pi^2 \hbar} \frac{eB}{\hbar} F(x),$$

where $\Delta \rho/\rho$ is the fractional change in resistivity due to the applied magnetic field, $\alpha$ is a coefficient between 0 and 1 that is thought to represent the Coulomb screening, $e$ is the elementary electronic charge, $\hbar$ is Planck’s constant, $B$ is the applied magnetic field, and $\rho$ is the resistivity. $F(x)$ is the Hurwitz zeta function given by [49]:

$$F(x) = \sum_{n=0}^{\infty} \left[2(n + 1 + 1/x)^{1/2} - 2(n + 1/x)^{1/2} - (n + 1/2 + 1/x)^{-1/2}\right],$$

where $x$ is a dimensionless quantity defined by $x = 4eD\tau^* B/\hbar$, $D$ is the electron diffusion constant, $\tau^*$ is the electron inelastic-scattering lifetime, and the phase coherence length $l_\phi = \sqrt{D\tau^*}$. In the high magnetic field regime (> 0.3 T), MC is dominated by the parabola-like $B^2$ dependence, originating from the Lorentzian deflection of carriers under perpendicular magnetic field. The electron executes cyclotron orbits, thereby shortening the mean free path, and thus decreasing the conductance [50]. We accounted for this by fitting the transport data. However, the fitting quality provides a worse description to experimental results, as exemplarily shown in figure 6(b) with the orange line for 0.42% Te. Using the HLN 2D model [4, 51, 52]:

$$\Delta G(B) = \frac{\beta e^2}{\pi \hbar} \left[ \Psi\left(\frac{1}{2} + \frac{B_x}{B}\right) - \ln \left(\frac{B_0}{B}\right) + 2 \ln \left(\frac{B_0 + B_{SO}}{B}\right) - 2\Psi\left(\frac{1}{2} + \frac{B_0 + B_{SO}}{B}\right) \right. - \left. \Psi\left(\frac{1}{2} + \frac{B_0 + 2B_{SO}}{B}\right) + \ln \left(\frac{B_0 + 2B_{SO}}{B}\right) \right],$$

(11)
Figure 6. Magneto-transport properties in heavily doped GaAs samples. (a)–(c) Show the MC measured at 3 K for GaAs:S, GaAs:Te and GaAs:Zn samples with different doping concentrations. Symbols are experimental points and solid lines are fitting results based on WL theory using the HLN model, equation (12). The orange line in figure (b) shows the 3D model fitting result for 0.42% Te. (d) and (f) Show the phase coherence length $l_\phi$ and the spin–orbit length $l_{so}$ at 3 K for $n$-type GaAs as a function of carrier concentration. (e) and (g) Show the temperature dependent phase coherence length $l_\phi$ and spin–orbit length $l_{so}$, respectively.

where $\Psi(x)$ is the digamma function of a physical variable $x$ [53] and $\beta$ is a constant of 1 (or $-1/2$) corresponding to WL (weak anti-localization). $B_\phi = \frac{\hbar}{4\beta e l_\phi}$ is a magnetic field defined by a phase coherence length $l_\phi$ ($l_\phi = \sqrt{D/\tau_{\phi}}$). $B_{so} = \frac{\hbar}{4\beta e l_{so}}$ is the characteristic magnetic field defined by the spin–orbit length $l_{so}$, which can lead to weak anti-localization effect. Since at high magnetic field the conductance is dominated by the large quadratic background ($B^2$ dependence) from the Lorentz force, the final fitting of the conductivity was performed using equation (12)

$$\Delta G(B)_{FIT} = \Delta G(B)_{HLN} + CB^2,$$

(12)
where $C$ is a measurement temperature related parameter [4], while in the fitting process of temperature range of 3–80 K, it keeps nearly constant.

The solid curves in figures 6(a) and (b) correspond to the fits with equation (12). The theory of WL in two-dimensional provides an excellent description of the transport results. Figure 6(c) shows the magneto-transport data obtained from $p$-type GaAs exhibiting the negative MC. This phenomenon is commonly explained by weak anti-localization originating from spin–orbit coupling (SOC), which has already been reported in heavily doped $p$-type Ge [4], and in some topological insulator materials [54]. However, by fitting the transport curve of the $p$-type GaAs using equation (12), we found that $l_0$ is negligibly small compared to $l_\phi$, which excludes the assumption that the negative MC in $p$-type GaAs can be attributed to the strong SOC. Here, the negative magnetoconductivity in $p$-type GaAs is attributed to the Lorentzian force instead of a weak-antilocalization.

Figures 6(d) and (f) show the phase coherence length and spin–orbit length as a function of carrier concentration at 3 K. As expected, the phase coherence length is comparable or larger (80–540 nm) than the thickness of the doped layer (100 nm), confirming that the system is quasi-two dimensional and justifies the use of HLN model.

The electron dephasing time $\tau_\phi$ corresponds to 0.5–38 ns with $l_\phi$ varying from 80 to 540 nm. The error bars for the phase coherence length and the spin–orbit length in figure 6 indicate the 95% confidence intervals of the fits. The 10% percent inaccuracy of the carrier concentration is due to the inaccuracy in determining the thickness of the doped layer. For the 0.07% S sample, we are unable to extract a proper phase coherence length and spin-diffusion length owing to the bad quality of the experimental data fittings, thus the discussion will focus on the rest of the $n$-type GaAs samples.

Figure 6(d) shows that $l_\phi$ increases from 80 to 540 nm with increasing the electron concentration from $5 \times 10^{18}$ to $5 \times 10^{19}$ cm$^{-3}$. Similarly, an enhancement of $l_\phi$ with increasing carrier concentration by applying back gating voltage is reported in low dimensional systems. Koester et al have reported that in Si/Sl$_{0.7}$Ge$_{0.3}$ quantum wells, at 1.3 K the phase coherence length can be tuned from 170 to 690 nm by increasing sheet electron concentration from $4.15 \times 10^{11}$ to $4.85 \times 10^{11}$ cm$^{-2}$ [55]. Premasiri et al have shown that for a multilayer indium monoselenide (InSe), at 2.2 K $l_\phi$ increases from 40 to 65 nm as the back gating voltage increases from 23 to 48 V, corresponding to an increase in sheet electron concentration from $3 \times 10^{12}$ to $13 \times 10^{12}$ cm$^{-2}$ [56]. The increase of $l_\phi$ is commonly attributed to the enhanced diffusion parameter and the increased Fermi velocity [57]. Figure 6(e) illustrates temperature dependent $l_\phi$, which decreases with increasing temperature and follows $l_\phi \propto T^n$ power law. The reported value $n = -1/2$ is ascribed to the dephasing mechanism being related to electron–electron collision with small energy transfer (or the ‘Nyquist dephasing’) in 2D systems [57, 58]. Since the electron–phonon scattering also plays a role in the inelastic scattering mechanism in heavily doped $n$-type GaAs, it could explain the value of $n$ between $-0.24$ and $-0.38$ as shown in figure 6(e). Moreover, the phase coherence length of about 540 nm for samples 0.42% Te and 0.32% S at 3 K could not fit with this power law, which may be related to the stronger screening of charged impurities at electron concentration reaching $5 \times 10^{19}$ cm$^{-3}$ [19]. It means electrons can obtain a longer phase memory before they are scattered by impurities, leading to an enhanced weak-localization for samples 0.42% Te and 0.32% S at 3 K. The $l_\phi$ as large as 540 nm at 3 K in this study is comparable with the values obtained in Si/Sl$_{0.7}$Ge$_{0.3}$ quantum well systems [55], which is desirable for quantum device applications.

Figure 6(f) show the extracted $l_\phi$ in the range of 29–48 nm for five heavily doped $n$-type GaAs samples. For Te doped GaAs, it shows a decreased $l_\phi$ with increasing the electron concentration. This is due to the reduced spin relaxation time $\tau_{so}$ since $l_\phi = \sqrt{D_{so}}$. Dzhioev et al have studied the spin relaxation time in GaAs by the optical orientation method [59]. It was found that $\tau_{so}$ decreases with increasing electron concentration in the metallic regime. With increasing the donor concentration from $10^{16}$ to $10^{19}$ cm$^{-3}$, the spin relaxation time decreases from 100 to 0.04 ns. Using $l_{so} = \sqrt{D_{so}}/\tau_{so}$, we have estimated $\tau_{so}$ in our Te-doped GaAs in the range of 0.06 and 0.18 ns. The decrease of $\tau_{so}$ was explained by Dyakonov–Perel (DP) spin dephasing mechanism [59]. It applies to materials without the inversion symmetry, like groups III–V (such as GaAs) and II–VI (ZnSe) compound semiconductors, where the inversion symmetry is broken by the presence of two distinct atoms in the Bravais lattice. However, in our case figure 6(g) illustrates that at temperature range of 3–80 K, $l_\phi$ for $n^{++}$ GaAs is insensitive to temperature, which suggests spin precession is due to carrier collisions with phonons or impurities named Elliott–Yafet (EY) spin relaxation mechanism instead of DP mechanism [60]. Kikkawa and Awschalom also reported that the EY mechanism dominates the spin precession process in $n$-type GaAs below 30 K [61]. For S doped GaAs, $l_{so}$ is generally larger at a similar electron concentration. Since heavier elements have stronger SOC, GaAs:Te gives rise to a shorter spin relaxation time than in GaAs:S. This explains the different behavior of $l_{so}$ for GaAs:Te and GaAs:S samples.
4. Conclusion

In summary, heavily doped GaAs samples synthesized by ion implantation followed by non-equilibrium annealing process are systematically investigated regarding their electrical and optical properties. We have shown that WL is responsible for the PMC in n-type GaAs samples. By fitting the magnetotransport data with the HLN model, the extracted phase coherence length and the spin–orbit length are as large as 540 and 48 nm, respectively, implying heavily doped GaAs as a potential material for quantum devices and spintronic application. Moreover, the unprecedented high electron concentration ($5 \times 10^{19} \text{ cm}^{-3}$) extends the possibilities of using GaAs as chemical sensors based on plasmonic effect.

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Data availability statement

All data that support the findings of this study are included within the article (and any supplementary files).

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