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Effect of high-temperature annealing on electrical and optical properties of undoped semi-insulating GaAs

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A comprehensive characterization, including room temperature Hall effect, near infrared absorption, temperature dependent dark current and photocurrent (using 1.13 eV light), normalized thermally stimulated current (NTSC), photoluminescence at 4.2 K in both near band edge and deep level regions, and selective pair photoluminescence (SPL) at 2 K, has been carried out on undoped semi-insulating GaAs samples, cut from four wafers which were grown by the low pressure liquid encapsulated Czochralski technique and annealed by three different schedules: a 1100 °C anneal with either fast or slow cooling, or a 1000 °C standard anneal. The 1100 °C anneal clearly introduces higher concentrations of NTSC traps near 0.3 and 0.5 eV, a PL center at 0.8 eV, and acceptor centers, which are mainly due to the point defects and increase the resistivity. Slow cooling to some extent reduces all of these additional centers. The SPL measurements show changes in the relative intensities of C, Zn, and Si related emissions with changes in annealing conditions.

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I. INTRODUCTION

To improve the microscopic resistivity and photoluminescence uniformity and to reduce As precipitates in undoped semi-insulating (SI) GaAs wafers, and thus attain a better uniformity and higher yield of GaAs devices, various annealing processes have been investigated by many different laboratories. Typical annealing schedules include: (1) single step ingot annealing at 900–1000 °C for homogenizing EL2 defects, which are mainly agglomerated along dislocation cell walls;¹ (2) multistep ingot annealing with, for example, two steps at 1100 °C and 950 °C, respectively, or three steps at 1100, 600, and 900 °C, respectively, for reducing As precipitates and diminishing deep centers other than EL2;² and (3) multistep wafer annealing, first at 1100 °C under As vapor pressure and then at 950 °C in an open furnace, for further reduction of As precipitates.³ The cooling rate following the high-temperature annealing has been found to play a key role in the ultimate properties of annealed SI GaAs materials. For example, a semiconducting/semi-insulating reversibility in undoped bulk GaAs was found to be controlled simply by the speed of the cooling rate after annealing at 950 °C, which indicates that point defects play an important role in the electrical compensation.⁴ In the present study, a comprehensive characterization, which includes room-temperature Hall effect, near-infrared absorption (NIRA) for [EL2] determination, temperature dependent dark current and photocurrent (TDDC) and (TDPC), normalized thermally stimulated current (NTSC), photoluminescence (PL) in both the near band edge and deep level regions, and selective pair PL (SPL) measurements, was performed on undoped SI GaAs boules grown by the low-pressure liquid-encapsulated Czochralski (LP-LEC) technique. Before being sliced into wafers, the boules were

subjected to high-temperature annealing at different temperatures and then were cooled at different rates. The particular annealing schedules included a 1100 °C anneal with either fast or slow cooling and a 1000 °C standard anneal. A significant variation of the electrical and optical properties in the annealed SI GaAs wafers was seen in nearly all of the measurements discussed above, and the results can be interpreted in terms of defect creation and changes in impurity incorporation.

II. SAMPLES AND EXPERIMENTS

A 4-in. LP-LEC SI GaAs boule was cut into three sections (A, B, and C) and annealed in various ways, as shown in Fig. 1. Section A, near the seed, was annealed at 1100 °C and cooled quickly (by pulling the ingot section from the furnace) and is labeled as 1100 °C-Q; Section B, in the middle of the boule, was also annealed at 1100 °C, but cooled slowly (by simply turning off the furnace) and is labeled as 1100 °C-A; and Section C, close to the tail, was annealed according to the Airtron 1000 °C standard anneal and is labeled as 1000 °C-Std. One wafer from each of Sections A and C (wafer 016A and wafer 001C), and two wafers from Section B (wafers 001B and 018B, to check for ingot uniformity and measurement consistency) were studied. To check the lateral uniformity within a wafer, center, ring, and edge pieces (6 mm×6 mm in size) from each wafer were measured and analyzed statistically. The metric for uniformity is the standard deviation (SD) of center, ring, and edge values divided by the mean, i.e., SD/mean, expressed as a percentage. Hall effect measurements were carried out at 296 K under vacuum and in the dark, using an automated high-impedance Hall effect apparatus with a magnetic-field strength of 4.8 kG. The neutral EL2 concentration was de-

FIG. 1. Wafer positions and annealing conditions in a 3 in. LP-LEC GaAs boule.

terminated by NIRA at $1.1 \mu\text{m}$, using a Ge photodiode as a detector. Deep centers in SI GaAs, acting as traps, can be studied by TSC spectroscopy. In the TSC measurements, the sample was prepared with alloyed In contacts and then cooled to a low temperature near 80 K, using a Bio-Rad DL4600 system, and finally illuminated with 1.45 eV light, which mainly causes electron excitation from neutral EL2. After a period of excitation long enough to completely fill all the electron traps, the illumination was removed and the sample was linearly heated at a rate β (e.g., 0.3 K/s) in the dark, resulting in a TSC spectrum. A TSC peak in the temperature spectrum can be correlated with the energy level for a given trap by the formula $E_T/kT_m = \ln(T_m^4/\beta) + \ln(1.7 \times 10^{16} \sigma_T/E_T)$, where T_m is the peak temperature, E_T and σ_T , the apparent energy level and capture cross section, respectively, and k , Boltzmann's constant.⁵ An Arrhenius plot of $\ln(T_m^4/\beta)$ vs $1/T_m$ can be used to obtain E_T and σ_T . To compare trap densities, a normalization procedure based on TDPC (using 1.13 eV light) was employed; it can be shown that the normalized ratio $I_{\text{TSC}}/I_{\text{PC}}$ (called NTSC) is independent of mobility μ , lifetime τ , and geometric factors.⁶ Thus the NTSC spectrum gives a truer picture of the relative trap densities. To study the dominant deep center, which controls the dark conductivity in undoped SI GaAs, the TDDC was also measured. Photoluminescence measurements were made

at 4.2 K using the 5145 Å line of an Ar laser as an excitation source. The PL signals were detected by using a cooled Ge detector or a photomultiplier tube with S-1 response. The signals were analyzed by a 3/4-m Spex spectrometer. SPL enables a definite identification of the residual shallow acceptors as well as a quantitative estimation of their relative concentrations.⁷ The measurement was performed at 2 K by using a 10 mW Ar ion laser pumped tunable dye laser. In this experiment the laser was tuned to 8200 Å. To search for the source of the acceptors, induced by high-temperature annealing, the residual impurities in four wafers were also analyzed by using glow discharge mass spectrometer (GDMS).

III. RESULTS AND DISCUSSION

The average electrical properties at 296 K, i.e., resistivity ρ , mobility μ , and electron concentration n , with their SD/mean values for four wafers, and the neutral, ionized, and total EL2 concentrations, i.e., $[\text{EL2}^0]$, $[\text{EL2}^+]$, and $[\text{EL2}]$, respectively, in the central portion of each wafer, are given in Table I. We first note that all of the results for wafer 018B, including Hall, NIRA, TSC, DC, and PC data, are very close to those of wafer 001B, which thus confirms the good vertical uniformity along an ingot section and also the precision and reliability of our measurements. From Table I, we see that: (1) a 1100 °C anneal (Sections A and B), as compared to the standard 1000 °C anneal (Section C), lowers the electron concentration and also the mobility; (2) fast cooling (after the 1100 °C anneal) results in a much lower carrier concentration than does slow cooling; and (3) the wafer with 1000 °C-Std. annealing shows the best uniformity in terms of SD/mean data for ρ , μ , and n , while the wafers with 1100 °C-Q and 1100 °C-A annealings have slightly poorer uniformities. For conductive samples, the lower carrier concentration and mobility were earlier explained by an addition of acceptors, but with SI materials, homogeneity may also play a role in the mobility. However, a detailed electrical compensation analysis based on the electron concentration, neutral EL2, TDDC, and TDPC data, as shown below, indeed indicates an increase of acceptors in wafers annealed at 1100 °C, especially those subjected to fast cooling. The question that we should answer concerns the possible sources of the acceptors.

TABLE I. Average 296-K electrical properties and EL2 concentration.

Wafer	Anneal		Resistivity $10^7 \Omega \text{ cm}$	Mobility $\text{cm}^2/\text{V s}$	Electron concentration 10^7 cm^{-3}	$[\text{EL2}^0]$	$[\text{EL2}^+]^a$	$[\text{EL2}]$
						in central portion		
						10^{16} cm^{-3}		
016A	1100 °C-Q	Mean	12.83	6237	0.79	1.02	0.37	1.39
		SD/mean	11%	3%	8%			
001B	1100 °C-A	Mean	4.68	6378	2.11	1.13	0.15	1.28
		SD/mean	11%	2%	10%			
018B	1100 °C-A	Mean	4.45	6693	2.10	1.22	0.155	1.38
		SD/mean	7%	4%	4%			
001C	1000 °C-Std	Mean	2.08	7397	4.07	1.41	0.09	1.50
		SD/mean	6%	1%	6%			

^a $[\text{EL2}^+]$ was estimated from Hall effect data, using $n = 2.6 \times 10^6 [\text{EL2}^0]/[\text{EL2}^+] \text{ cm}^{-3}$.

FIG. 2. Arrhenius plots of temperature dependent dark current and photocurrent (using 1.13 eV light) for the center samples cut from four wafers.

Arrhenius plots of TDDC and TDPC, using 1.13 eV light, for the center samples from four wafers, are presented in Fig. 2. Again, the results for the two samples of Section B are very close to each other. From Fig. 2, we also see that: (1) the activation energy for each DC plot is 0.77 eV, showing the control of EL2 in the dark conductivity; (2) the average activation energy for the PC plots at $T > 140$ K is about 82 meV, which is very close to the thermal activation energy of the electron capture cross section for EL2 (66 meV),⁸ if the temperature dependence of electron mobility is considered; (3) a clear IR quenching of PC at $T > 140$ K is found in sample 016A (1100 °C-Q), which is due to the IR quenching of EL2 and is probably associated with the high concentration of ionized EL2, i.e., $EL2^+$; and (4) the PC's are proportional to the DC's at $T > 200$ K. From observation 1, we infer that both carrier concentration and DC for all of the samples (assuming nearly equal mobilities) should be proportional to $N_{EL2}/N_A - 1$, where N_{EL2} is the total EL2 concentration and N_A , the net acceptor concentration. Because $N_{EL2} = N_{EL2}^0 + N_{EL2}^+$ and $N_A = N_{EL2}^+$ (since $n \ll N_A$), n and DC are also proportional to N_{EL2}^0/N_{EL2}^+ . From the n , $[EL2^0]$, and estimated $[EL2^+]$, data in Table I, or the 300-K DC data in Fig. 2, it is easily found that as compared to sample 001C (1000 °C-Std.), the net acceptor concentration in sample 016A (1100 °C-Q) is increased by a factor of 4 and those in samples 001B and 018B (1100 °C-A), by a factor of 1.7. The possible point defect nature for the high-temperature annealing induced acceptors will be discussed later. As stated in observation 4, based on the data obtained from all the samples used in the study, a linear relationship between DC and PC at 300 K is found, as plotted in Fig. 3, which implies that the lifetime of electrons photoexcited from the $EL2^0$ center is inversely proportional to the $EL2^+$ concentration, i.e., the higher the $EL2^+$ or N_A concentration, the lower the carrier lifetime.⁶ From the ratio of PC/DC at 300 K, which is about 2 for the light intensity of 3.3×10^{14} phot/cm² s used in this study, an electron capture cross section of

FIG. 3. Linear relationship between dark current and photocurrent at 300 K for 12 different samples cut from the center, ring, and edge of four wafers.

1.4×10^{-16} cm² at 300 K can be deduced for the $EL2^+$ center.⁹

NTSC spectra for the four center samples are given in Fig. 4. From the figure, several significant effects of the annealing and cooling conditions can be observed: (1) the ratio of T_2 over T_3 is strongly affected, changing from greater than unity for sample 001C (1000 °C-Std.) to less than unity for sample 016A (1100 °C-Q); (2) T_5 is increased by the 1100 °C-A annealing, as compared to that of the 1000 °C-Std. annealing; and (3) more T_5^* appears, and some T_5 is lost in sample 016A (1100 °C-Q), resulting in an overlap of the two traps. Once again, the two samples from Section B show a very similar spectral structure. Our previous studies^{6,10,11} indicate that: (1) T_2 , with $E_T = 0.63$ eV and $\sigma_T = 1.9 \times 10^{-13}$ cm², is related to As_{Ga} , but is not identical to $EL2$ (also related to As_{Ga}) because of its different activation energy; (2) T_3 , with $E_T = 0.50$ eV and $\sigma_T = 1.1 \times 10^{-13}$ cm², might be associated with V_{As} , most likely as a V_{As} -related defect complex; and (3) T_5 , with $E_T = 0.35$ eV and σ_T

FIG. 4. Normalized TSC spectra for center samples cut from four wafers.

FIG. 5. Near-band-edge PL spectrum (4.2 K) of the center samples cut from four wafers (note each curve with its own scale).

$\approx 1.9 \times 10^{-14} \text{ cm}^2$, is very similar to EL6 and might be a complex involving both As_{Ga} and V_{As} or $V_{\text{Ga}}-V_{\text{As}}$ and As_i .^{12,13} From an analysis of the density of T_2 in various undoped SI GaAs samples,⁶ we know that it is roughly proportional to $[\text{EL}2^0]$. In the present study, $[\text{EL}2^0]$ changes only about 30% among the four wafers. Thus, the main effect of the annealing at 1100 °C, in comparison with the 1000 °C standard annealing, is to increase the V_{As} - and V_{Ga} -related point defects.

The near band edge PL spectra from the center samples of the four wafers are shown in Fig. 5 (note each curve with its own scale). The transition for the decay of a free exciton (FE), decay of an exciton bound to a neutral acceptor (A^0, X), and annihilation of a donor electron and acceptor hole (D, A) due to C_{As} are shown. For the 001C sample (1000 °C-Std.) the (D, A) transition is at a slightly lower energy than in the other three (1100 °C annealed) samples. Also, the 001C sample shows an additional transition at lower energy than the transition marked (D, A) and is indicative of multiple acceptors. The presence of multiple acceptors in this sample will be further discussed in terms of selective pair PL. One of the striking differences is the magnitude of the PL signal. The intensities of the (D, A) transition for 001C, 001B, 018B, and 016A are 75.0, 3.0, 4.2, and 0.39 mV, respectively, as shown in the Fig. 5. Thus, the intensity of the 001C (1000 °C-Std.) sample is 20 times greater than the average intensity of the (1100 °C-A) samples, and the average intensity of the (1100 °C-A) samples is 9.3 times higher than the 016A (1100 °C-Q) intensity. On the low energy side of (A^0, X), 001B shows three distinct sharp transitions, 018B and 016A each show two, while 001C shows only the slightest hint of a transition. These sharp lines are attributed to bound exciton recombination at defect pairs, preferentially oriented parallel to one of the $\langle 110 \rangle$ directions in the (001) growth plane.¹⁴ Figure 6 shows the PL spectra from the center samples of the four wafers for energies from above the band gap down to ap-

FIG. 6. Deep-level-region PL spectra (4.2 K) for the center samples cut from four wafers (note all curves with a same scale).

proximately a third of the band gap (note all curves have the same scale). The discontinuity at 0.775 eV is caused by a filter change to maintain purity of the PL beam. The presence of the EL2-related As_{Ga} defect is indicated by the two overlapping emission bands at 0.64 and 0.68 eV, respectively, attributed to conduction band to As_{Ga} , and As_{Ga} to valence band recombinations.¹⁵ The 0.8 eV emission band has been correlated with the EL6 center seen in DLTS.¹⁶ We believe the presence of this 0.8-eV radiative recombination center effectively shunts the near band-gap (D, A) transition and accounts for the factor of 20 decrease between the 001C (1000 °C-Std.) intensity and the average of the (1100 °C-A) intensities, and the further 9.3 decrease between the average of the (1100 °C-A) intensities and the 016A (1100 °C-Q) intensity. Control of the near-band-gap PL intensity by the recombination center responsible for the 0.8 eV band has also been reported by Alt *et al.*¹⁷ in a room temperature PL mapping study of SI GaAs, followed by low-temperature PL measurements of the below band-gap region. A characteristic of the Ga antisite (Ga_{As}) or $\text{Ga}_{\text{As}}-V_{\text{Ga}}$ is an emission transition at 1.441 eV (Ref. 18) or 1.45 eV.⁴ However, none of the four samples in this study show the presence of the Ga_{As} defect in their PL spectrum of this region (not shown).

Figure 7 presents a comparison of 2 K SPL spectra from wafers 016A (1100 °C-Q), 001B (1100 °C-A), and 001C (1000 °C-Std.). The spectrum from wafer 018B (not shown) is similar to that of wafer 001B. From the figure we see that: (1) four transitions due to $C:2S_{3/2}$, $\text{Zn}:2S_{3/2}$, $\text{Zn}:2P_{5/2}$ (Γ_8), and $\text{Si}:2S_{3/2}$ can be identified in all three wafers; (2) unexpectedly, $C:2S_{3/2}$ is not a dominant SPL emission in these wafers; and (3) the dominant SPL emission in the wafers with either 1100 °C-Q or 1100 °C-A annealing is from zinc (Zn_{Ga}), while in the wafer with 1000 °C-Std. annealing, is from silicon (Si_{As}). The change of dominant SPL emission from Zn_{Ga} to Si_{As} caused by the anneal condition implies that the incorporation of impurities, like Zn and Si, depends on the availability of Ga and As vacancies, which might be closely related to the anneal temperatures and the cooling

FIG. 7. 2 K selective pair PL spectra for the center samples cut from three wafers.

conditions. An M-shape distribution of $\text{Zn:}2\text{S}_{3/2}$ across an as-grown SI GaAs wafer, as opposed to the usual W-shape distribution of EL2 or dislocation density, was reported by Koteles *et al.*⁷ Since EL2 is a As_{Ga} -related defect complex and the Zn acceptor occupies the Ga site, the M shape of Zn_{Ga} versus the W shape of As_{Ga} across an as-grown SI GaAs wafer means that a competition might exist between As and Zn for V_{Ga} sites or an anticorrelation between As_{Ga} and V_{Ga} across the diameter of the wafer.

The introduction of acceptors into various GaAs materials, as a function of annealing and cooling conditions have been extensively studied in the past decade.^{4,19,20} The thermal conversion of SI GaAs from *n* to *p* type by high-temperature annealing at 1000–1200 °C, either under As vapor pressure of 1 atm.³ or in vacuum²¹ followed by fast cooling, has also been reported. Based on the thermodynamic equilibria of point defects, including antisites, vacancies, and interstitials of Ga and As, at the crystal growth temperature, and their interactions during the cooling process, many studies have concluded that intrinsic point defects, such as Ga_{As} or $\text{Ga}_{\text{As}}-V_{\text{Ga}}$, are responsible for the acceptor. However, to explain the *n*- to *p*-type thermal conversion and the reduction of EL2 concentration by high-temperature annealing at $T > 1150$ °C, an As-antisite breakdown mechanism has been also proposed; i.e., the As_{Ga} breaks into As_i and V_{Ga} ,²¹ and the latter can be a multiply charged deep acceptor ($E_v + 0.52$ eV).²² In the present study, the near band edge PL spectra reveal an absence of Ga_{As} -related defects, possible candidates for the acceptors, in all four wafers, while the SPL spectra show a more dominating $\text{Zn:}2\text{S}_{3/2}$ transition due to Zn_{Ga} in the wafer after the 1100 °C-Q annealing. The results imply that in the wafer with 1100 °C-Q annealing there exist more V_{Ga} sites, which assists the incorporation of Zn impurity, as compared to the wafer with 1000 °C-Std. annealing. GDMS results show that the dominant acceptor in the four wafers is carbon, with a maximum change of a factor of two in their atomic densities, and that the other two acceptors are silicon and zinc, but with much lower atomic densities. That is, the Si concentration

FIG. 8. Regular TSC spectra for samples 001C (1000 °C-Std) and 016A (1100 °C-Q).

varies less than a factor two, from 17 ppb in sample 016A, to 9 ppb in samples 001B and 018B, and to 20 ppb in sample 001C. Zn varies from 22 ppb in sample 016A, to 17 ppb in sample 001B and 31 ppb in sample 018B, and to 30 ppb in sample 001C. These results cannot account for the variation of the acceptor concentration found in samples 016A and 001C, which is estimated to be a factor of four. However, the variation in the PL intensity observed in the four samples can be correlated with the 0.8 eV band and NTSC of T_5 , both are considered to be related to a defect complex involving V_{Ga} , V_{As} , and As_i . Therefore we infer that intrinsic point defects, such as V_{Ga} , contribute to the introduction of acceptors into the wafer with 1100 °C-Q annealing. The other possible source of acceptors is the contaminant Cu, entering the samples during high-temperature annealing.²³ In the regular TSC spectra of samples 001C (1000 °C-Std.) and 016A (1100 °C-Q), as shown in Fig. 8, a TSC trap (T_a at 170 K), which has an activation energy of 0.43 eV and has been correlated with a 4.2 K, Cu-related PL emission at 1.36 eV and a Cu-related DLTS center at $E_v + 0.42$ eV^{24–26} is indeed observed in the sample 016A (1100 °C-Q).

To understand the semiconducting/semi-insulating reversibility of GaAs caused by a 950 °C soak followed by slow or fast cooling, a tentative model was proposed by one of the present authors and his co-workers.⁴ In this model, a donor (*D*) at $E_c - 0.13$ eV, observed in the semiconducting material, is attributed to $V_{\text{As}}-\text{As}_{\text{Ga}}$, and an acceptor (*A*) at $E_v + 0.07$ eV, observed in the semi-insulating material, is assigned to $V_{\text{Ga}}-\text{Ga}_{\text{As}}$. A rapid cooldown (quench) freezes in significant amounts (10^{15} – 10^{16} cm⁻³) of $V_{\text{As}}-\text{As}_{\text{Ga}}$ and $V_{\text{Ga}}-\text{Ga}_{\text{As}}$, and the latter complexes (plus other acceptors below midgap) dominate the former complexes (plus other donors above midgap) so that the deep donor EL2 can render the sample semi-insulating ($[A] > [D]$ and $[\text{EL2}] > [A] - [D]$). A slow cooldown (anneal), on the other hand, permits the rather unstable $V_{\text{Ga}}-\text{Ga}_{\text{As}}$ to breakup as E_F rises, so that $[D] > [A]$ and the sample is conductive. However, it is also known that some donors shallower than EL2 can be created in significant amounts (2×10^{15} cm⁻³) by heat treat-

ment at 470 °C for a relatively long time,²⁰ and these could also raise the Fermi level of the slowly cooled wafer, increasing the electron concentration. In the present study, we find a clear contrast in the electrical and optical properties between the wafers with 1100 °C-Q and 1100 °C-A anneals, but no resistivity conversion, as observed in the previously mentioned 950 °C investigation.⁴ In comparison with 1100 °C-A annealing, 1100 °C-Q annealing causes: (1) a significant decrease of electron concentration, DC, and PC due to an increase of acceptor concentration, which also results in a decrease of carrier lifetime since $N_A = [EL2^+]$; (2) an obvious increase of NTSC traps T_3 and T_5^* (both of which are probably related to the As vacancy); (3) a clear increase of the 0.8 eV band in the PL spectrum, which is believed to correlate with a DLTS center, EL6;¹⁶ and (4) more incorporation of carbon (C:2S_{3/2}) and zinc [Zn:2P_{5/2}(Γ₈)] impurities in the SPL spectrum. Basically, all these results can be explained by using the point defect model proposed in Ref. 4. In the 1100 °C-Q annealing, we infer that: (1) the V_{Ga}-related defect centers dominate the V_{As}-As_{Ga} centers and contribute an excess of acceptors; and (2) the V_{As}-As_{Ga} or V_{As}-As_i-V_{Ga} complexes contribute deep donor centers, like T_3 and T_5^* in NTSC and the 0.8 eV band in PL. In the 1100 °C-A annealing, on the other hand, the point complexes mentioned above can be annihilated by interactions with Ga and As interstitials, resulting in a decrease of the point defect related acceptors and deep donors. Unlike the case for the 950 °C-A sample studied in Ref. 4, the *dominant* deep center in the wafer with 1100 °C-A annealing is a 0.77 eV center due to EL2, which means that the V_{As}-As_{Ga} complex at $E_C - 0.13$ eV, even if it has reformed during some slow cooling stage at 400–600 °C, does not exist in a sufficient concentration to compensate the acceptors induced by high-temperature annealing at 1100 °C.

Although the high-temperature annealing followed by fast cooling produces higher-resistivity and lower-mobility SI GaAs material, the electrical and optical uniformities of the material are quite good. This technique should be explored further, especially by comparing the performance of GaAs devices and integrated circuits fabricated on the various materials.

IV. CONCLUSIONS

The effects of high-temperature annealing at 1100 °C, with fast or slow cooling, on the electrical and optical properties of undoped LP-LEC SI-GaAs have been studied by Hall effect, NIRA, TDDC, TDPC, NTSC, PL, and SPL measurements. In comparison with the Airtron 1000 °C standard annealing, the 1100 °C anneal clearly introduces: (1) higher concentrations of NTSC traps, T_5 near 0.3 eV and T_3 near 0.5 eV, both probably V_{As} or V_{As}-V_{Ga} related; (2) higher intensities of the 4.2-K PL emission band at 0.8 eV along with a significant decrease in the PL intensity of the (D,A) transition, due to a decrease of carrier lifetime; and (3) higher concentrations of acceptor centers, dominated by point defects, like V_{Ga}, rather than by shallow impurities, resulting in a significant increase of the resistivity. The slow

cooling reduces all of these additional centers to some extent. The SPL measurements at 2 K show a clear change in the relative intensities of C_{As}, Zn_{Ga}, and Si_{As} related emissions with changes in annealing conditions.

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- ¹ R. Rumsby, I. Grant, M. R. Brozel, E. J. Foulkes, and R. M. Ware, in *Semi-Insulating III-V Materials*, edited by D. C. Look and J. S. Blakemore (Shiva, Nantwich, 1984), p. 165.
- ² K. Kashiwa, Y. Otoki, M. Seki, S. Taharasako, and S. Okubo, *Hitachi Cable Rev.* **9**, 55 (1990); Y. Otoki, M. Watanabe, T. Inada, and S. Kuma, *J. Cryst. Growth* **103**, 85 (1990).
- ³ M. Mori, G. Kano, T. Inoue, H. Shimakura, H. Yamamoto, and O. Oda, in *Semi-Insulating III-V Materials*, edited by A. G. Milnes and C. J. Miner (Adam Hilger, New York, 1990), p. 59.
- ⁴ D. C. Look, P. W. Yu, W. M. Theis, W. Ford, G. Mathur, J. R. Sizelove, D. H. Lee, and S. S. Li, *Appl. Phys. Lett.* **49**, 1083 (1986).
- ⁵ D. C. Look, *Semicond. Semimet.* **19**, 75 (1983).
- ⁶ Z.-Q. Fang and D. C. Look, *Inst. Phys. Conf. Ser.* **149**, 43 (1996).
- ⁷ E. S. Koteles, J. Kafalas, S. Zemon, and P. Norris, *Inst. Phys. Conf. Ser.* **83**, 63 (1987).
- ⁸ A. Mitonneau, A. Mircea, G. M. Martin, and D. Pons, *Rev. Phys. Appl.* **14**, 853 (1979).
- ⁹ D. C. Look and Z.-Q. Fang, *J. Appl. Phys.* **80**, 3590 (1996).
- ¹⁰ Z.-Q. Fang, D. C. Look, S. Kuisma, K. Saarinen, and P. Hautajarvi, in *Proceedings of 9th Conference on Semiconducting and Insulating Materials, Toulouse, France*, edited by C. Fontaine (IEEE, New York, 1996), p. 149.
- ¹¹ Z.-Q. Fang, D. C. Look, H. Yamamoto, and H. Shimakura, *Appl. Phys. Lett.* **69**, 3417 (1996).
- ¹² T. Wosinski, A. Makosat, and Z. Witzak, *Semicond. Sci. Technol.* **9**, 2047 (1994).
- ¹³ Z.-Q. Fang, T. E. Schlesinger, and A. G. Milnes, *J. Appl. Phys.* **61**, 5047 (1987).
- ¹⁴ M. S. Skolnick, T. D. Harris, C. W. Tu, T. M. Brennan, and M. D. Sturge, *Appl. Phys. Lett.* **46**, 427 (1985).
- ¹⁵ M. Tajima, *Jpn. J. Appl. Phys., Part 1* **26**, L1060 (1987).
- ¹⁶ S. Alaya, M. A. Zaidi, G. Marrakchi, H. Maaref, H. J. von Bardeleben, and J. C. Bourgoin, *Mater. Res. Soc. Symp. Proc.* **163**, 163 (1990).
- ¹⁷ H. Ch. Alt, M. Mullenborn, and G. Packeiser, in *Semi-Insulating III-V Materials*, edited by A. G. Milnes and C. J. Miner (Adam Hilger, New York, 1990), p. 309.
- ¹⁸ S. Alaya, H. Maaref, H. J. von Bardeleben, and J. C. Bourgoin, *Appl. Phys. Lett.* **56**, 1877 (1990).
- ¹⁹ T. Asom, J. M. Parsey, Jr., L. C. Kimerling, R. Sauer, and F. A. Thiel, *Appl. Phys. Lett.* **52**, 1472 (1988).
- ²⁰ Y. Kitagawara, N. Noto, T. Takahashi, and T. Takenaka, *Appl. Phys. Lett.* **52**, 221 (1988).
- ²¹ N. Ohkubo, M. Shishikura, and S. Matsumoto, *J. Appl. Phys.* **73**, 615 (1993).
- ²² H. Xu and U. Lindefelt, *Phys. Rev. B* **41**, 5979 (1990).
- ²³ T. Hiramoto and T. Ikoma, in *Semi-Insulating III-V Materials*, Malmö, Sweden, edited by G. Grossmann and L. Ledebø (Adam Hilger, Bristol and Philadelphia, 1988), p. 337.
- ²⁴ Z.-Q. Fang, D. C. Look, and R. L. Jones, *J. Electron. Mater.* **26**, L29 (1997).
- ²⁵ Z. G. Wang, H. P. Gislason, and B. Monemar, *J. Appl. Phys.* **58**, 230 (1985).
- ²⁶ A. Mitonneau, G. M. Martin, and A. Mircea, *Electron. Lett.* **13**, 666 (1977).