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# Compensation Mechanism in Liquid Encapsulated Czochralski GaAs: Importance of Melt Stoichiometry

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**Abstract**—It is shown that the key to reproducible growth of undoped semi-insulating GaAs by the liquid encapsulated Czochralski (LEC) technique is the control over the melt stoichiometry. Twelve crystals were grown from stoichiometric and nonstoichiometric melts. The material was characterized by secondary ion mass spectrometry, localized vibrational mode far infrared spectroscopy, Hall-effect measurements, optical absorption, and photoluminescence. A quantitative model for the compensation mechanism in the semi-insulating material was developed based on these measurements. The free carrier concentration is controlled by the balance between EL2 deep donors and carbon acceptors; furthermore, the incorporation of EL2 is controlled by the melt stoichiometry, increasing as the As atom fraction in the melt increases. As a result, semi-insulating material can be grown only from melts above a critical As composition. The practical significance of these results is discussed in terms of achieving high yield and reproducibility in the crystal growth process.

## I. INTRODUCTION

A MAJOR impediment to the development of GaAs integrated circuits and microwave devices has been the lack of a reliable supply of semi-insulating substrates with reproducible and thermally stable [1] properties suitable

for device fabrication. The liquid encapsulated Czochralski (LEC) technique is currently receiving considerable attention for the production of GaAs crystals because semi-insulating material can be grown without intentional [2] doping, and the technique offers the potential for producing round, large-area substrates with uniform properties at a reasonable cost [3], [4].

The main question surrounding semi-insulating LEC GaAs has concerned the compensation mechanism by which the undoped material is semi-insulating. The understanding of the compensation mechanism has two important practical consequences. First, knowledge of the cause-effect relationships between crystal growth and electrical characteristics of the material would greatly improve the yield of semi-insulating crystals in the growth process, as well as the crystal-to-crystal and wafer-to-wafer reproducibility. Second, this understanding would improve device performance. For example, backgating effects could possibly be diminished by adjusting [5] trap levels in material intended for integrated circuit processing.

It has been shown that the melt stoichiometry [6] is the key parameter which must be controlled to grow undoped semi-insulating LEC GaAs. This finding has led to the development of quantitative model for the compensation mechanism. The purpose of this paper is to give a detailed account of these experimental results and interpretations. In the first section of the paper it is shown that the

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resistivity is controlled by the melt stoichiometry. The second section describes the identification of the predominant chemical impurities and electrically active centers in the semi-insulating material. The compensation model is developed in the third section.

## II. APPARATUS AND PROCEDURES

The crystals used in this study were grown in the  $\langle 100 \rangle$  direction from pyrolytic boron nitride (PBN) crucibles using a "Melbourn" high pressure (Metals Research, Ltd.) crystal puller. The pressure of the argon ambient during growth was 300 lb/in<sup>2</sup>. All of the crystals were undoped. The crystal weight and diameter were typically 2.5 kg and 3 in, respectively. Nominally dry ( $< 500$  ppm H<sub>2</sub>O) B<sub>2</sub>O<sub>3</sub> encapsulating material (Johnson Matthey) was used for all growth experiments. The melt was prepared by *in situ* [7] synthesis, to achieve high purity, starting with a charge of 6–9's As (Cominco) and 6–9's Ga (Kawecki). The composition of the melt was varied by changing the relative amounts of Ga and As in the initial charge. The composition of the initial melt was determined by adjusting the charge composition for the amount of As lost [6] during the heat-up cycle. The weight of lost As was determined by comparing the weight of the initial charge to the weight of the crystal and the charge remaining in the crucible after growth. The B<sub>2</sub>O<sub>3</sub> was carefully separated from the charge remaining in the crucible before weighing to eliminate possible errors introduced by B<sub>2</sub>O<sub>3</sub> evaporation. Samples for characterization were obtained from along the length of each crystal. The melt composition corresponding to the growth of each sample was determined by adjusting the initial melt composition for the crystal weight at the time of growth. The crystal weight and length were recorded during the growth as a function of time. After terminating the growth process, the crystals were cooled at between 30 and 80°C h<sup>-1</sup>.

The concentration of background transition metal and group III–VI impurities was determined by secondary ion mass spectrometry (SIMS) [8]. The concentration of carbon (acceptors) was determined by localized vibrational mode (LVM) far-infrared spectroscopy [9]. The LVM absorption band for <sup>12</sup>C occurs at 582 cm<sup>-1</sup>. The integrated absorption is proportional to the <sup>12</sup>C concentration. The carbon concentration of our material was determined by measuring the absorption at 77 K and converting to concentration by using the optical cross section given in [9]. The free carrier concentration, Hall mobility, and resistivity were determined by Van der Pauw–Hall-effect measurements at room temperature and over a temperature range of from 298 to 400 K. Optical absorption measurements in the near infrared were made at room temperature with a Perkin–Elmer Model 330 spectrometer. Photoluminescence measurements [10] were made at 4.2 K by immersing samples in liquid helium.

## III. STOICHIOMETRY-CONTROLLED RESISTIVITY

The electrical characteristics of the crystals were evaluated by Hall-effect measurements using samples obtained from the fronts and tails of the twelve crystals, and from

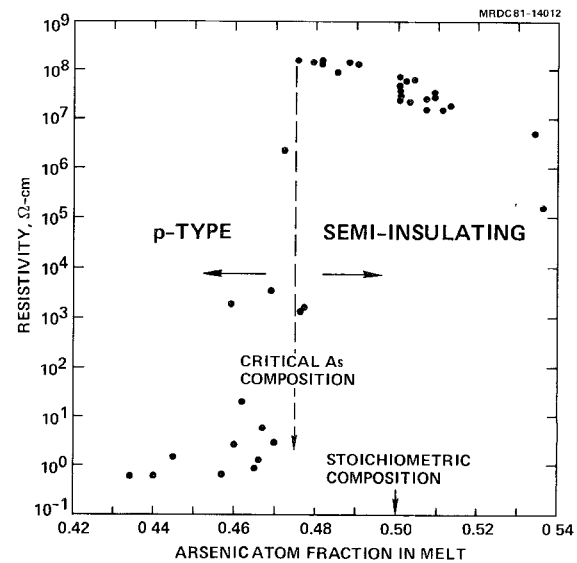


Fig. 1. Dependence of electrical resistivity of LEC GaAs on melt stoichiometry. Semi-insulating material is obtained above, and *p*-type (low-resistivity) material below a critical melt composition of about 0.475 As atom fraction.

detailed resistivity profiles of five crystals. The resistivity was found to be a strong function of the melt stoichiometry, as shown in Fig. 1.

Fig. 1 shows that the material is semi-insulating (*n*-type) above, and *p*-type (low resistivity) below a critical As concentration in the melt of about 0.475 As atom fraction. The resistivity peaks at the critical composition at a value of about  $1.5 \times 10^8 \Omega \cdot \text{cm}$  and decreases approximately eight orders of magnitude below the critical composition. The resistivity also decreases very gradually as the As fraction increases from the critical composition.

The variation in resistivity across the melt composition range is explained in terms of the corresponding free carrier concentration and Hall mobility, as shown in Figs. 2 and 3. The semi-insulating material grown at the critical composition is *n*-type with a carrier concentration and mobility of  $1\text{--}2 \times 10^7 \text{ cm}^{-3}$  and  $1\text{--}2 \times 10^3 \text{ cm}^2 \text{ V}^{-1} \text{ s}^{-1}$ , respectively. These Hall mobilities are low for *n*-type GaAs, possibly indicating the presence of inhomogeneities in the material. As the As atom fraction increases from the critical composition to about 0.51, the mobility gradually increases to  $4\text{--}5 \times 10^3 \text{ cm}^2 \text{ V}^{-1} \text{ s}^{-1}$ , which is more typical of *n*-type material. The corresponding electron concentration gradually increases to  $6\text{--}8 \times 10^7 \text{ cm}^{-3}$ . The combined increase of both the mobility and carrier concentration lead to a reduction in resistivity of about one order of magnitude. A further change in the As fraction from 0.51 to about 0.54 brings about a more pronounced increase in free carrier concentration of about two orders of magnitude. The result is a corresponding decrease in resistivity to as low as  $1.5 \times 10^5 \Omega \cdot \text{cm}$ .

The material becomes *p*-type below the critical composition. The free hole concentration rises approximately nine orders of magnitude following a 1 percent reduction in As fraction in the melt from the critical composition. The hole concentration and Hall mobility of this material are in the range of  $1\text{--}3 \times 10^{16} \text{ cm}^{-3}$  and  $215\text{--}330 \text{ cm}^2 \text{ V}^{-1} \text{ s}^{-1}$ ,

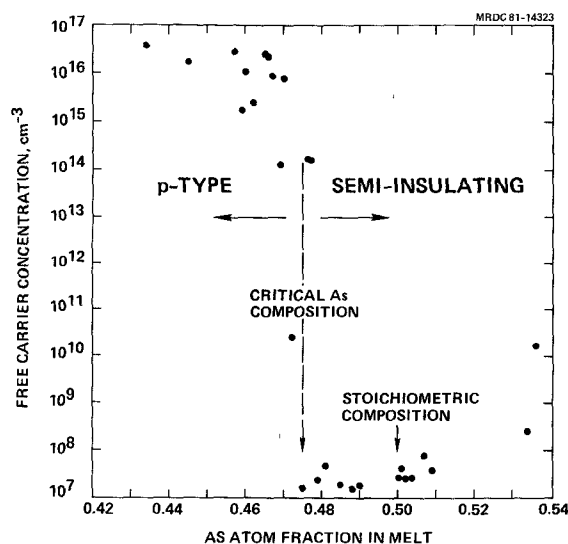


Fig. 2. Dependence of free carrier concentration of LEC GaAs on melt stoichiometry. The semi-insulating material is *n*-type, and the free electron concentration increases gradually as the As concentration in the melt increases from the critical composition. The material becomes *p*-type below the critical composition. The free carrier concentration rises approximately nine orders of magnitude following a 1 percent reduction in As fraction from the critical composition.

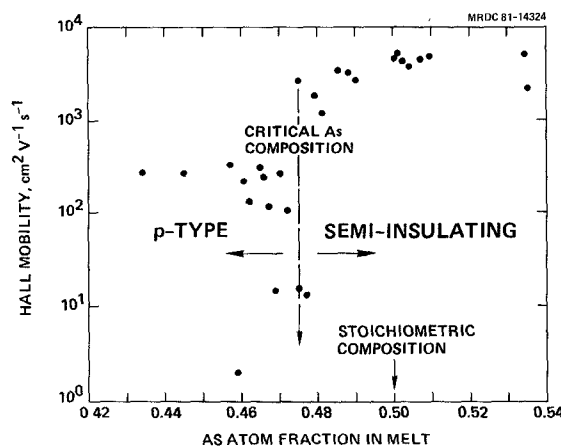


Fig. 3. Dependence of Hall mobility of LEC GaAs on melt stoichiometry. The mobility of the semi-insulating material varies from  $1-2 \times 10^3$  to  $4-5 \times 10^3$   $\text{cm}^2 \text{V}^{-1} \text{s}^{-1}$  as the As atom fraction increases from the critical composition to about 0.535. The mobility of the *p*-type material grown in the transition region, within about 1 percent of the critical composition, is low, between  $1-30$   $\text{cm}^2 \text{V}^{-1} \text{s}^{-1}$ . The mobility of the *p*-type material grown outside of the transition region ranges from 215 to  $330$   $\text{cm}^2 \text{V}^{-1} \text{s}^{-1}$ .

respectively. Mobilities obtained from the *p*-type material grown in the transition region, corresponding to melt compositions within about 1 percent of the critical composition, were very low, between  $1$  and  $30$   $\text{cm}^2 \text{V}^{-1} \text{s}^{-1}$ . The measured hole concentrations were between  $1 \times 10^{10}$  and  $2 \times 10^{14} \text{cm}^{-3}$ . These carrier concentrations are too high to explain the low mobilities in terms of mixed conduction. The low mobilities of material grown in the transition region could reflect inhomogeneities in the material. For instance, a striated pattern of regions of high and low resistivity could cause such behavior.

Detailed resistivity profiles of crystals grown from initially As- and Ga-rich melts further emphasize the role of the melt stoichiometry in controlling the electrical com-

pensation. It is important to note that unless the initial melt is precisely stoichiometric [11], As-rich (Ga-rich) melts become progressively more As-rich (Ga-rich) as the crystal is pulled from the melt. Crystals grown from As-rich melts were invariably semi-insulating from front to tail. Crystals grown from Ga-rich melts initially below the critical composition were *p*-type throughout. On the other hand, crystals grown from Ga-rich melts initially above the critical composition underwent a transition from semi-insulating to *p*-type at the point along the crystal where the corresponding melt composition reached the critical value. This behavior clearly indicates that the resistivity is controlled by the melt stoichiometry, and, that the semi-insulating-to-*p*-type transition is not related to the normal segregation of some common background impurity toward the tail of the crystal. Otherwise, the tail of As-rich-grown crystals would have become *p*-type as well.

#### IV. ANALYSIS OF CHEMICAL IMPURITIES AND ELECTRICALLY ACTIVE CENTERS

To establish a model for the compensation mechanism in undoped semi-insulating LEC GaAs, the chemical impurities and electrically active centers were evaluated. The chemical impurities, as determined by SIMS and LVM, are shown in Table I. These results represent an average of measurements obtained from the front and tail of each of the twelve crystals. A range of concentrations is reported for those elements which exhibited substantial concentration variations from crystal to crystal. Carbon, as determined by LVM, is the predominant, chemically identifiable, electrically active background impurity. The concentration of carbon ranged from  $2 \times 10^{15}$  to  $1.3 \times 10^{16} \text{cm}^{-3}$ . The concentration of the other background impurities common to GaAs, such as S, Mg, Cr, Mn, and Fe, is very low. In particular, the Si concentration is consistently less than about  $1 \times 10^{15} \text{cm}^{-3}$  as determined by SIMS measurements. Low Si contamination results from crystal growth without a quartz crucible. Si levels in this material are more than one order of magnitude lower than in Bridgman material that was analyzed. Boron is present in a range from about  $1 \times 10^{16}$  to  $6 \times 10^{17} \text{cm}^{-3}$ . The source of boron contamination is the  $\text{B}_2\text{O}_3$  encapsulant. The incorporation of boron has been shown [12] to depend on the water content of the encapsulant, decreasing as the water content increases. Although boron is the predominant chemical impurity, it is iso-electronic with Ga, and no evidence was found in this study indicating that boron is electrically active.

The evaluation of the electrical and optical properties of the semi-insulating material indicates that the deep donor, commonly referred to as EL2, is the predominant deep center. An optical absorption band between  $1$  and  $1.4 \mu\text{m}$  previously identified with the EL2 center [13] was observed in all of the semi-insulating material. In addition, the activation energy of the electron concentration, obtained from plots of the temperature-corrected free electron concentration as a function of the reciprocal of temperature, was  $0.75 \pm 0.02 \text{eV}$ . This energy is consistent with published values [14] for the activation energy of EL2. The behavior

TABLE I  
CHEMICAL IMPURITY ANALYSIS OF 12 LEC GaAs CRYSTALS

| AVERAGE*<br>CONCENTRATION<br>(cm <sup>-3</sup> ) | IMPURITY |       |       |      |       |       |       |       |       |         |           |
|--|----------|-------|-------|------|-------|-------|-------|-------|-------|---------|-----------|
|  | S        | Se    | Te    | Mg   | Cr    | Mn    | Fe    | Cu    | Si    | B       | C #       |
|  | 1 1e15   | <5e14 | <5e13 | 2e14 | <5e14 | <1e15 | <3e15 | <9e14 | <1e15 | 1-60e16 | 2 1-13e15 |

\* 12 CRYSTALS ANALYZED

# CARBON DETERMINED BY LVM

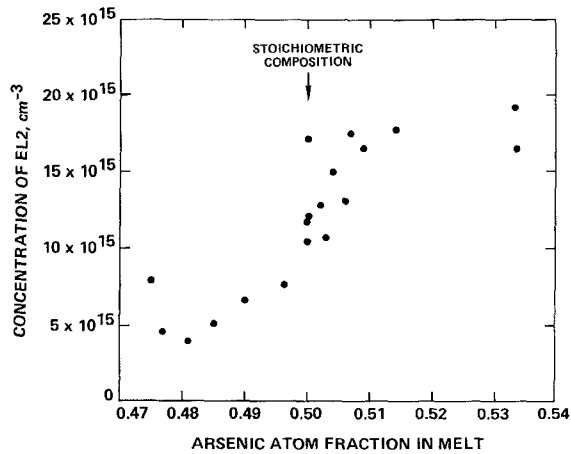


Fig. 4. Dependence of EL2 concentration as determined by optical absorption on the melt stoichiometry. The concentration of EL2 increases from  $5 \times 10^{15} \text{ cm}^{-3}$  to  $1.7 \times 10^{16} \text{ cm}^{-3}$  as the As atom fraction increases from about 0.48 to 0.51, and appears to saturate as the As concentration increases further to 0.53.

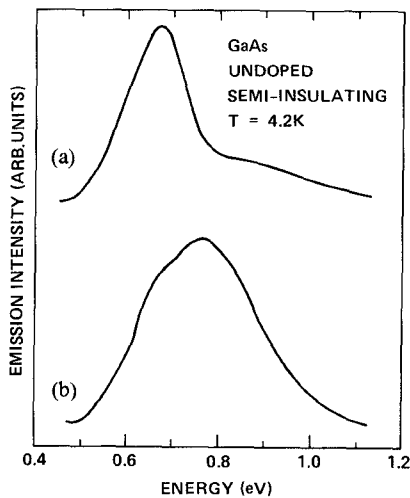


Fig. 5. Typical photoluminescence spectra of semi-insulating GaAs grown from (a) an As-rich melt (As atom fraction = 0.507,  $\rho = 1.8 \times 10^7 \Omega \cdot \text{cm}$ ) and (b) a Ga-rich melt (As atom fraction = 0.488,  $\rho = 1.4 \times 10^8 \Omega \cdot \text{cm}$ ). The intensity of the 0.68-eV band decreases as the As atom fraction decreases toward the critical composition consistent with the optical absorption measurements (Fig. 4).

of the photoconductivity thresholds [15] above and below 120 K was also found to be consistent with the presence of EL2.

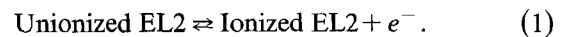
The concentration of EL2 was determined by optical absorption using the cross section reported by Martin *et al.* [16]. Absorption due to unoccupied EL2 centers was not observed, and the Hall measurements indicated that the

center was more than 90 percent occupied. Consequently, the absorption was taken to be proportional to the total EL2 concentration. The concentration of EL2 was found to depend on the melt stoichiometry, as shown in Fig. 4, increasing from about  $5 \times 10^{15}$  to  $1.7 \times 10^{16} \text{ cm}^{-3}$  as the As atom fraction increased from about 0.48 to 0.51. The concentration remained constant as the As fraction increased further to about 0.535.

Photoluminescence (PL) studies of the semi-insulating material are consistent with the measured dependence of EL2 on melt stoichiometry. Typical PL spectra of semi-insulating material grown from Ga- and As-rich melts, shown in Fig. 5, exhibit bands peaking at 0.68 and 0.77 eV. The 0.68-eV band has been attributed [10] to radiative-recombination between EL2 electron traps and the valence band, and the 0.77-eV band to recombination possibly associated [10] with a hole trap. The intensity of the 0.68-eV band in the semi-insulating GaAs grown from Ga-rich melts is substantially reduced by comparison with As-rich grown material. This behavior is consistent with the decrease of the EL2 concentration with decreasing As fraction (Fig. 4) as determined by optical absorption. Neither band was observed in the *p*-type material.

## V. THE COMPENSATION MECHANISM

To develop a model for the electrical compensation in terms of the concentration of a predominant electrically active centers in the semi-insulating material, the concentration of shallow and deep centers were related through first-principle theoretical considerations. The ionization of EL2 produces an ionized center plus an electron in the conduction band



According to the principle of detailed balance, the concentration of ionized centers  $N_I$ , the concentration of electrons  $n$ , and the concentration of unionized centers  $N_U$  are related by the following equation:

$$\frac{N_I n}{N_U} = K \quad (2)$$

where  $K$  is a constant determined by the thermodynamics of the system.  $N_I$  is equal to the net acceptor concentration, given as the difference in concentration between shallow acceptors  $N_A$  and shallow donors  $N_D$

$$N_I = N_A - N_D \quad (3)$$

The concentration of acceptors is given as the sum of the

concentrations of carbon and other residual acceptors  $N_A^R$

$$N_A = [\text{carbon}] + N_A^R. \quad (4)$$

The concentration of unionized centers is equal to the EL2 concentration as determined by optical absorption. That is, only EL2 centers which are occupied by electrons contribute to the optical absorption process

$$N_U = \text{EL2}. \quad (5)$$

By substituting (3) through (5) into (2), the following expression for the free electron concentration is obtained in terms of the predominant centers in the material:

$$n = K \frac{\text{EL2}}{([\text{carbon}] + N_A^R - N_D)}. \quad (6)$$

This expression can be rewritten in the following form:

$$[\text{carbon}] = K \frac{\text{EL2}}{n} + N_D - N_A^R. \quad (7)$$

Therefore, the carbon concentration is proportional to the ratio of the EL2 concentration to the electron concentration.

Semi-insulating material grown from melts ranging from 0.475 to 0.535 As atom fraction was evaluated according to (7) from measurements of the carbon concentration (by LVM), the EL2 concentration (by optical absorption), and the electron concentration (by Hall-effect measurements) for each sample. A plot of the carbon concentration as a function of the ratio of the EL2 concentration to the electron concentration, shown in Fig. 6, follows linear behavior, indicating that the electron concentration is indeed controlled by the balance between EL2 and carbon. This result is independent of possible errors in the published values of the optical cross sections for carbon and EL2. It is important to note that if some other impurity were the predominant acceptor, such as Mn, Fe, Cu, or Zn, the linearity predicted on the basis of (7) would still necessarily hold. However, the linearity would not be distinguishable because the term  $[\text{carbon}]$  would be small compared to  $N_A^R$ ; the figure would be a scatter plot. In fact, the scatter in these data probably reflects actual fluctuations in the concentration of other background impurities rather than random error in the experimental measurements. The small value of the intercept ( $N_D - N_A^R$ ) of the least-squares fit to the data also indicates the predominance of carbon acceptors. Thus, EL2 deep donors and carbon acceptors control the electrical compensation in semi-insulating LEC GaAs grown from melts ranging from 0.475 to 0.535 As atom fraction.

The variation of the electrical characteristics of the semi-insulating material (Figs. 1–3) with melt stoichiometry can now be explained on the basis of the analysis above. The EL2 concentration must either exactly match or exceed the carbon concentration to produce semi-insulating properties. The EL2 concentration in material grown from Ga-rich melts below the critical composition is insufficient to compensate the carbon, leading to *p*-type conductivity. Semi-insulating material grown at the critical com-

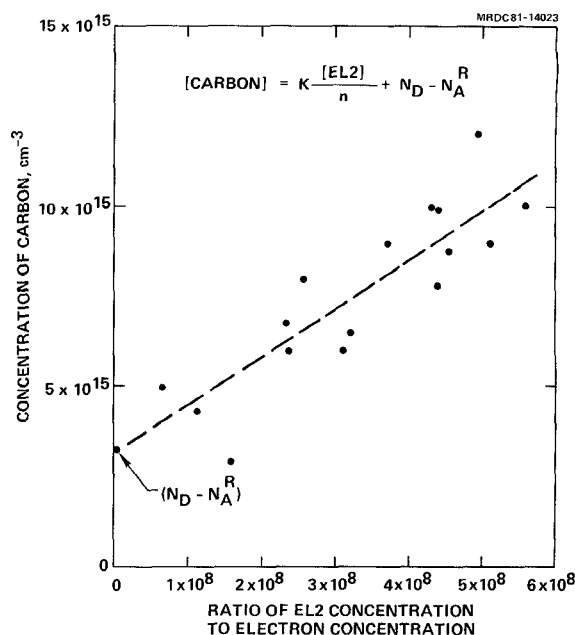


Fig. 6. Dependence of the carbon concentration on the ratio of the EL2 concentration to the electron concentration. The concentration of carbon, EL2, and electrons was determined for each sample. The dashed line is a least-square fit to the data. The linearity of the data indicates the dominant roles played by EL2 deep donors and carbon acceptors in controlling the compensation (see text). The small value of the intercept ( $N_D - N_A^R$ ) also indicates the predominance of carbon acceptors.

position is closely compensated, leading to the maxima of the resistivity. As the As atom fraction in the melt increases from the critical composition to about 0.51, the EL2 concentration becomes progressively higher than the carbon concentration. As a result, thermal ionization of (unionized) EL2 centers (see (1)) gives rise to a gradual increase in the electron concentration and a corresponding decrease in the resistivity (Figs. 1 and 2). Our model further shows that the relatively low resistivity of semi-insulating material grown from melts between 0.53 and 0.54 As atom fraction is due to exceptionally low concentrations of carbon acceptors.

## VI. SUMMARY AND CONCLUSIONS

It was shown that the melt stoichiometry controls the compensation in undoped GaAs grown by the liquid encapsulated Czochralski technique. The free carrier concentration of the semi-insulating material is controlled by the balance between EL2 donors and carbon acceptors. The incorporation of EL2 depends on the melt stoichiometry, increasing as the As melt concentration increases. As a result, semi-insulating material can be grown only above a critical As concentration.

From a fundamental standpoint, these results have provided the first direct evidence indicating that the identity of the EL2 center is associated with a native defect or native-defect complex in bulk GaAs: LEC GaAs is pure enough to reveal the contribution of native-defects in bulk material. The dependence of EL2 on melt stoichiometry is consistent with published work on GaAs grown by vapor-phase epitaxy [17] and organometallic chemical-vapor de-

position [18]. Isolated native defects which would follow the stoichiometry dependence of EL2 (Fig. 4) include the gallium vacancy  $V_{Ga}$ , the arsenic interstitial  $As_i$ , and the antisite  $As_{Ga}$ . Since  $V_{Ga}$  would be expected to be an acceptor, EL2 would more likely be related to one of the latter two defects. There is evidence [10], [19]–[24] supporting either defect, and we believe that further work on well-characterized material is needed to make a definitive assignment of EL2.

In practical terms, these results show that semi-insulating GaAs can be grown by the LEC technique reproducibly and with high yield provided that the melt is As-rich. This condition ensures that the melt will not become Ga-rich during the growth process. Indeed, nine crystals grown during the course of this investigation from near-stoichiometric As-rich melts were semi-insulating from front to tail. In addition, the characteristics of Se active layers implanted directly into the substrates showed excellent uniformity [25] of the pinchoff voltage across each wafer. These results demonstrate that high yield and reproducibility can be achieved in both the growth of GaAs by the LEC technique and in the application of LEC GaAs to the integrated circuit technology.

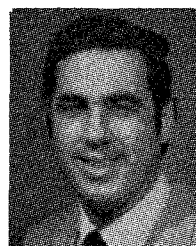
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Dr. Holmes is a member of the American Association of Crystal Growth, and an associate member of Sigma Xi.

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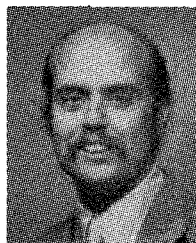


R. T. Chen (M'76) was born in Taiwan on April 11, 1950. He received the B.S. degree in physics from the National Central University, Chung-Li, Taiwan, in 1972, and the Ph.D. degree in materials science from the University of Southern California, Los Angeles, CA, in 1980. His doctoral dissertation focused on the defect structure of heavily Si-doped GaAs and its influence on physical properties. During his graduate studies he was also engaged in GaAs crystal growth and Cr electrodeposition.

In April 1980 he joined Rockwell International Microelectronics Research and Development Center as a Member of the Technical Staff, and he is currently working on improving the liquid encapsulated Czochralski (LEC) crystal growth technique for high purity, large diameter, semi-insulating and degenerately doped GaAs substrate material. He is the coauthor of 12 scientific papers.

Dr. Chen is a member of the New York Academy of Sciences, the Physical Society, and the Electrochemical Society.

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Kenneth R. Elliott attended Purdue University from 1970 to 1974, and obtained the B.S. degree in physics. From 1974 to 1979 he studied at the California Institute of Technology receiving the Ph.D. degree in applied physics. His thesis work concentrated on the properties of impurities and defects in semiconductors as determined by the photoluminescence, optical absorption, and excitation spectroscopy of excitons bound to impurity sites. In 1980, he assumed a post-graduate position at the Max-Planck-Institut Hochfeld-Magnetlabor (High Magnetic Field Laboratory) in Grenoble, France, where his work concentrated on the photoluminescence of electron irradiated silicon.

Since November 1980 he has been a Member of the Technical Staff at Rockwell International's Microelectronics Research and Development Center. His work at Rockwell has concentrated on deep-level studies of semi-insulating GaAs and investigations of interface states in GaAs MIS structures. He is the author of more than 10 publications in the field of semiconductor physics.

Dr. Elliott is a member of Phi Beta Kappa.

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**C. G. Kirkpatrick** (S'69-S'74-M'74-SM'79) received the B.S. and A.M. degrees in physics in 1969 and 1970, respectively, from Washington University, and the Ph.D. degree in electrical engineering from the University of Illinois in 1974. While at Washington University, she worked on programs which involved the development of techniques for studying radiation damage through track etching and microscopy in meteorites and lunar samples, and studies of possible nonlinearities in the photoelectric effect

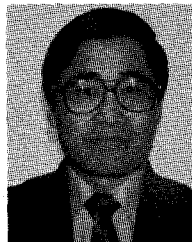
at very low light intensities. Her thesis work at the University of Illinois was a study of radiation damage in silicon resulting from ion and electron irradiation by means of photo-luminescence and electrical measurements. She held the Zonta International Amelia Earhart Fellowship during 1972-1973.

In 1974 she joined the Corporate Research and Development Center of the General Electric Company. Her research in that laboratory centered on beam (ion, electron, and laser) effects in semiconductors studied through a variety of analytical techniques, especially electron microscopy (EBIC). The development of beam-addressed memory devices utilized the findings of these studies. She holds five patents in this area. Her work at General Electric also entailed other applications of beams for imaging and display devices, electron lithography and microscopy, ion implantation,

and laser and electron beam annealing. In 1979 she joined Rockwell Science Center, where her work has included the study of ion implantation effects in GaAs, and the investigation of beam annealing and alloying in GaAs. She presently directs the activities of members of the Materials Technology Section, whose efforts are centered on ion implantation, encapsulation, bulk growth, and characterization of GaAs. She has authored or coauthored more than 30 publications in the areas of ion and electron beam effects and semiconductor materials.

Dr. Kirkpatrick is a senior member of the Society of Women Engineers, and a member of the American Physical Society.

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**Phil Won Yu** was born in Yonan, Korea. He received the B.S. degree in nuclear engineering from Inha Institute of Technology, Incheon, Korea, in 1961, the M.S. in physics from Yonsei University, Seoul, Korea, in 1964 and the Ph.D. in physics from Brown University, Providence, RI, in 1971.

From 1964 to 1966 he was a full-time Instructor in physics at Yonsei University Seoul, Korea. From 1971 to 1972 he was a Research Assistant at Brown University, Providence, RI, working on the electron-hole plasma in Ge. In 1972 he joined the University of Dayton, Dayton, OH as a Research Physicist working in the areas of electrical and optical characterizations of III-V, II-VI, and I-III-VI<sub>2</sub> compounds, LED and photodetectors, and ion implementation technology. In 1980 he joined Wright State University as a Senior Research Physicist. His current interests include the deep center properties of GaAs, InP, and the related compounds and their microwave and optoelectronic applications. He has authored or coauthored about 40 publications.

## Proton Isolation for GaAs Integrated Circuits

DONALD C. D'AVANZO, MEMBER, IEEE

**Abstract**—Significant improvement in the electrical isolation of closely spaced GaAs integrated circuit (IC) devices has been achieved with proton implantation. Isolation voltages have been increased by a factor of four in comparison to a selective implant process. In addition, the tendency of negatively biased ohmic contacts to reduce the current flow in neighboring MESFET's (backgating) has been reduced by at least a factor of three. The GaAs IC compatible process includes implantation of protons through the SiO<sub>2</sub> field oxide and a three-layered dielectric-Au mask which is definable

to 3- $\mu$ m linewidths and is easily removed. High temperature storage tests have demonstrated that proton isolation, with lifetimes on the order of 10<sup>5</sup> h at 290°C, is not a lifetime limiting component in a GaAs IC process.

### I. INTRODUCTION

**T**HE SUCCESSFUL design and fabrication of medium- and large-scale GaAs integrated circuits (IC's) requires a high degree of electrical isolation between closely spaced active devices. Traditionally, mesa etched [1], [2] and selective implant [3] processes have been used to isolate GaAs IC's. These processes can result in significant current flow

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