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Publication Date
1983-04-01
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April 1983
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THE NATURE OF THE DOMINANT \(\gamma\)-INDUCED DEFECTS IN HIGH-PURITY GERMANIUM*

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*This work was supported in part by the Director's Office of Energy Research, Office of Health and Environmental Research, U.S. Department of Energy under Contract No. DE-AC03-76SF00098, and in part by the U.S. National Science Foundation under Contract No. DMR-8203430.
THE NATURE OF THE DOMINANT $\gamma$-INDUCED DEFECTS IN HIGH-PURITY GERMANIUM

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ABSTRACT

For relatively low Co-60 $\gamma$-irradiation doses ($3 \times 10^{15} \, \gamma \, \text{cm}^{-2}$) and dose rates ($2.0 \times 10^{10} \, \gamma \, \text{cm}^{-2} \, \text{sec}^{-1}$), two hole traps ($E_v + 0.23 \, \text{eV}, E_v + 0.38 \, \text{eV}$) are the dominant defects introduced in high-purity p-type germanium. A comprehensive study involving crystals grown under a wide variety of conditions supports the hypothesis that these centers are most likely due to complexes between oxygen and lattice vacancies. Injection-enhanced annealing was observed for both these hole-trapping states, and a discussion is given of the seemingly dose-dependent nature of the defects produced in $\gamma$-irradiated high-purity p-type germanium.

INTRODUCTION

The study of the defects induced in germanium by $\gamma$-irradiation has a long history¹. As in other semiconductors, the electrical and optical properties of germanium are sensitive to the introduction of the relatively stable lattice imperfections created by $\gamma$-irradiation. While many experimental results have been accumulated on the energy levels, introduction rates, and annealing behavior of $\gamma$-induced defects, a clear understanding of their chemical nature has
not been achieved in most cases. The radiation damage centers produced are extremely sensitive to the experimental conditions during the irradiation and to the type and density of background impurity levels. In intentionally doped n-type germanium, several levels ascribed to dopant-vacancy complexes and divacancies are observed. In p-type germanium, the picture is less clear, particularly in ultra high-purity material.

Many shallow and semi-deep levels in high-purity germanium have been identified as containing combinations of the four common electrically inactive contaminants—hydrogen, oxygen, carbon and silicon. In material quenched from high temperatures (875°C), all of the resulting deep level defects were correlated with the presence of oxygen. In γ- or e-irradiated silicon, the A-center (oxygen-vacancy complex) is one of the most prominent levels, and several oxygen-related defects have been identified in heat-treated silicon. Bearing these results in mind, we have attempted to identify the chemical nature of the dominant γ-induced defects in high-purity p-type germanium by conducting an extensive survey of the type and concentration of defect states created in crystals grown under a wide variety of conditions. A preliminary report of this work has appeared previously.

EXPERIMENTAL

A large number of crystals from various suppliers were used in this investigation. Table I shows a summary of the growth conditions of the 32 crystals used, which were obtained from five different sources: Lawrence Berkeley Laboratory, Australian Atomic Energy Commission, General Electric Co., Hoboken-Metallurgie and Sylvania Electric Incorporated. All except four of the crystals were grown by the Czochralski technique under varying
atmospheres. (The normal method of producing high-purity germanium is to pull the crystal from a melt contained in a silica crucible under a reducing atmosphere of pure H₂). For the Czochralski method, it is instructive to note the relative amounts of hydrogen, oxygen, silicon or carbon incorporated in the crystal. A typical crystal grown from a silicon crucible under a H₂ atmosphere may contain ~ 6 x 10¹³ oxygen atoms cm⁻³, while use of a N₂ atmosphere may increase this to ~ 10¹⁴ cm⁻³. Growth from a graphite crucible under H₂ may reduce the oxygen content to ~ 5 x 10¹² cm⁻³, as determined by the lithium precipitation method. The concentration of silicon incorporated may be determined by Spark Source Mass Spectrometry, and is typically 10¹⁴ – 10¹⁷ cm⁻³, the high end of this range being due to deliberate doping of the melt by addition of silicon. Crystals grown under a H₂ atmosphere also typically contain ~ 2 x 10¹⁵ hydrogen atoms cm⁻³, while crystals grown out of graphite crucibles may also contain 10¹⁴ – 10¹⁵ carbon atoms cm⁻³ as determined by the self-counting radiation detector technique. The other impurities always present are the electrically active shallow level acceptor impurities, generally boron, gallium or aluminum (with phosphorus as a minority impurity). These control the electrical purity of the material, and may be reduced to the 10¹⁰ cm⁻³ level.

Samples were fabricated into diode structures (~ 5 x 5 x 2 cm⁻³) using a variety of contacting methods. In general, lithium diffused n⁺ contacts (10 min, 300°C) were used; however, both antimony diffused (1 hr, 600°C) and ³¹p implanted (25 keV, 10¹⁴ cm⁻², damage annealed at 330°C) n⁺ contacts were also used at times during the course of the investigation. Evaporated palladium or implanted ¹¹B (25 keV, 10¹⁴ cm⁻², not annealed) served as ohmic (p⁺) contacts. Irradiations were performed at room temperature in either a fuel rod cooling pond facility (3 x 10¹⁰ γ cm⁻² sec⁻¹) or a standard ⁶⁰Co irradiation source (2 x 10¹⁰ γ cm⁻² sec⁻¹).
The primary method of characterizing the resulting defects was deep level transient spectroscopy (DLTS), performed in systems based on an electronic correlator and a 1 MHz capacitance bridge (either Boonton 71A or 72B*). This technique has the advantage of spectroscopically displaying the defects present in reverse biased diode structures at a high sensitivity. Defect concentrations were obtained from the magnitude of the capacitance transients caused by each individual defect, combined with the measurement of the net electrically active acceptor concentration, as determined by 1 MHz C - V measurements.

RESULTS

Nature of the γ-Induced Levels

It became obvious after a small number of samples that two deep hole trapping centers were predominant in γ-irradiated HP p-type germanium. Figure 1 shows the DLTS spectrum of a high-purity p-type sample (AAEC material) after a dose of $2.8 \times 10^{15} \, \gamma \, \text{cm}^{-2}$. The two main defects evident are measured to have energy levels of $E_v + 0.23 \, \text{eV}$ and $E_v + 0.38 \, \text{eV}$ respectively. These were determined from the slope of the Arrhenius plots in Fig. 2, after accounting for the $T^2$ dependence of the $<v_h>N_v$ product in the hole thermal emission rate expression:

$$e_h = \frac{\sigma_h <v_h>N_v}{g} \exp \left( \frac{-\Delta E}{kT} \right) \quad (1)$$

where $e_h$ is the thermal emission rate of captured holes at temperature $T$, $\sigma_h$ is the capture cross section for holes, $<v_h>$ is the average hole thermal

*Reference to a company or product name does not imply approval or recommendation of the product by the University of California or the U.S. Department of Energy to the exclusion of others that may be suitable.
velocity, $N_v$ is the density of states in the valence band, $g$ is a degeneracy factor (assumed = 2), $\Delta E$ is the defect energy level and $k$ is Boltzmann's constant.

The two hole traps were present in all normal high-purity material to a greater or lesser extent, as evidenced by the DLTS spectra from LBL and GE samples (Figs. 3 and 4). The presence of copper-related defects deliberately introduced prior to irradiation by diffusion from a surface source did not appear to affect the introduction of the two hole traps, nor did the irradiation have any effect on the copper-related centers (Fig. 5).

The capture cross-sections for holes of the two centers were measured using the bias pulse width dependence of the DLTS peak height, and were found to be temperature dependent according to the relations:

$$E_v + 0.23 \text{ eV}: \sigma = 5.9 \times 10^{-8} T^{-3.4} \text{ cm}^2 \quad (2)$$

$$E_v + 0.38 \text{ eV}: \sigma = 2.4 \times 10^{-8} T^{-2.9} \text{ cm}^2 \quad (3)$$

We have previously noted that a cascade capture model involving excited states could account for these temperature dependences. The dependence of relative DLTS signal versus bias pulse width at a fixed temperature for the two centers is shown in Fig. 6.

The presence of lithium was found to suppress formation of the two centers. Figures 7(a) and (b) show the relative damage caused in a lithium-drifted sample compared with an undrifted companion sample. The regions investigated were identical (~ 80 µm from the lithium-diffused contact), and the lithium-drifted sample shows far less radiation damage. The suppression of the defects by lithium was observed many times on the "tail" of lithium-diffused contacts on p-type γ-irradiated samples.
We have previously reported the effects of heat treating high-purity p-type germanium for various periods (1 - 1000 hours) at various temperatures (500 - 675°C) prior to irradiation\textsuperscript{16}. Such samples were considerably hardened to radiation damage compared with their unannealed companions, even though net doping densities had not significantly changed. Annealing under H\textsubscript{2} or N\textsubscript{2} appeared to make no significant difference, indicating that hydrogen is not likely to be a component of either of the two acceptor centers, as long-term annealing under N\textsubscript{2} ambients leads to the outdiffusion of hydrogen\textsuperscript{17}. As well, irradiated samples from crystals grown under N\textsubscript{2} or vacuum also showed the two deep levels at comparable concentrations to those grown under a H\textsubscript{2} atmosphere, provided that silica crucibles were used for all growths. The long-term room temperature stability of the two centers was also investigated, and it was found that the defects had greater stability in silica-grown material than in graphite-grown material\textsuperscript{15}. The major difference between these two classes of material is in the higher oxygen content of the silica-grown crystals. Cleland\textsuperscript{18} has also observed stabilizing of γ-induced defects in material of higher oxygen content.

No difference in concentration of the two hole traps was noticed in irradiated material of widely varying (0.008 - 3 μg g\textsuperscript{-1}) silicon content, provided all other conditions were constant; thus silicon does not appear to be part of the defects either.

Reduced concentrations of the two defect states were found in several "special" crystals. Figure 8(a) shows the relative damage created in two high-purity p-type samples from crystals grown under a H\textsubscript{2} atmosphere, one pulled from a graphite susceptor, the other from a quartz crucible. The major difference between these two crystals (apart from carbon content, which is not related to the chemical nature of the two centers) is in the lower oxygen content of the graphite-grown material, as discussed earlier. The lower defect
concentration in this material compared to the silica-grown crystal points to oxygen being part of the hole-trapping states. This is confirmed by the result of Fig. 8(b) which shows the DLTS spectrum of an irradiated aluminum-doped AAEC crystal grown from a graphite crucible under H₂. There is no evidence of the two defect states, even after a dose of \(5 \times 10^{16} \gamma \text{cm}^{-2}\). This sample had an extremely low oxygen content as determined by the rate of lithium precipitation; a normal lithium-diffused contact (10 min, 350°C) precipitated completely in less than an hour as determined by a thermal probe used for measuring the conduction type of germanium samples. By comparison, lithium contacts prepared on samples from crystals grown from silica crucibles are stable for periods of years at room temperature. Consequently, these low oxygen-content samples were prepared for irradiation with antimony-diffused (one hour, 500°C) contacts. Even allowing for possible radiation hardening effects, the two hole trapping states should be evident at concentrations of \(~1\%\) of the net background doping density; instead there is no evidence for them down to the sensitivity of the DLTS technique (<10⁻⁴).

The second class of material not to show these levels was the deliberately zinc-doped material obtained from Sylvania Electric. These crystals had been grown specifically for lithium drift compensation, and therefore had a low oxygen content.

The third class of material not to show these levels was normal grade lithium-driftable germanium from Hoboken. Sections from one of these crystals had previously shown normal lithium drift rates, and therefore low oxygen content. One of four sections from this crystal not previously drifted did show a very low concentration of the \(E_v + 0.23\) eV and \(E_v + 0.38\) eV centers after irradiation \((N_T \sim 3 \times 10^{-4}\) of the net background doping density). Figures 9(a) and (b) show DLTS spectra from this and one of its companion sections, which showed no evidence of the levels. For the dose received
(6 x 10^{15} \gamma \text{ cm}^{-2})$ the defects should have been present at concentrations of several per cent of the background doping density judging from the results on silica-grown material. Lithium contacts to these samples proved stable for the three days required for irradiation.

**Thermal and Injection-Enhanced Annealing**

As mentioned earlier, the thermal stability of the two hole traps had been investigated in a previous report\textsuperscript{15}. Figure 10 shows the result of a series of 20 min. isochronal heat treatments in $H_2$ for $\gamma$-irradiated samples containing the $E_v + 0.23$ eV and $E_v + 0.38$ eV centers. The two defect states follow the same annealing curve, and the growth conditions appeared to make no significant difference.

The stability of the levels under injection conditions was also examined by applying various forward biases for different times as a function of temperature. Figure 11 displays the 290 K isothermal annealing curve of the two defect levels under varying injection current ($40 - 200$ mA/cm$^2$), and shows that annealing is faster at higher injection levels. The slope of these annealing curves is shown as a function of injection current density in Fig. 12, and shows that we did not reach the saturation injection condition at these low values of forward current density. The defect levels are known to be stable under reverse bias at these temperatures, and indeed some defect motion over a long period under the action of a high electric field has been observed\textsuperscript{19}. Thus it is likely that a recombination-enhanced mechanism is responsible for the observed annealing of the two-hole trapping states under injection conditions, rather than some charge state mechanism. Injection-enhanced annealing is common in radiation damage centers in compound semiconductors\textsuperscript{20-22}, and has been observed previously in an electron-irradiation induced donor level in $n$-type germanium\textsuperscript{3}. 
DISCUSSION

The most significant features of the above results may be examined as follows. First, hydrogen and silicon do not appear to be components of either of the deep level hole trapping centers. Second, carbon does not appear to be involved, as the levels are normally more prominent in silica-grown material. Third, high-purity samples ($N_A - N_D < 5 \times 10^{10} \text{cm}^{-3}$) which were over-irradiated so that the net carrier density at 77 K rose by at least several orders of magnitude ($N_A - N_D \sim 2 \times 10^{12} \text{cm}^{-3}$) due to the introduction of deep radiation-induced acceptors showed concentrations of the $E_V + 0.23 \text{ eV}$ and $E_V + 0.38 \text{ eV}$ centers ($\sim 10^{12} \text{cm}^{-3}$) far in excess of the net impurity concentration at 77 K before irradiation. This means that the shallow level impurities are unlikely to be components of the two defect states.

This leaves oxygen as the likely component of the $\gamma$-induced centers, in a complex with vacancies. The four crystals with known low oxygen content showed either no evidence of the deep hole traps, or else an abnormally low concentration of them. All silica-grown crystals will have significant oxygen content (reaction between Ge, H$_2$ and SiO$_2$), and even the pyrocarbon-coated silica crucibles used for some of the LBL crystals could enable enough oxygen to be incorporated into the melt during crystal growth to produce significant concentrations of the two-hole traps on irradiation (the reaction of carbon and silica is a source of oxygen). There is always difficulty determining the exact amount of oxygen present in high-purity crystals because of the relatively small amounts (compared to, for example, silicon crystals), and the complexities of oxygen-complex formation after heat treatments. It is known that the form of oxygen in germanium can change during heating and, for example, be incorporated into electrically active shallow donor or acceptor centers.
Attention is drawn to the observation that the $E_V + 0.23 \text{ eV}$ center displays sensitivity in some samples to infra-red illumination from a GaAs light emitting diode (Fig. 13). By creating minority carriers available for capture, it may be determine how the two hole trapping states are populated. The $E_V + 0.38 \text{ eV}$ level does not show this sensitivity which may indicate that the $E_V + 0.23 \text{ eV}$ level has a relatively larger electron capture cross section. Shallow level defects containing oxygen are also sensitive to infra-red illumination\textsuperscript{24}.

The reduced concentration of the two hole traps in the presence of lithium is most likely due to the capture of oxygen into stable lithium-oxygen pairs as well as a much lesser component caused by direct passivation of the induced levels by the lithium, as with some laser-induced defects\textsuperscript{25}.

The introduction rates of the two defect states are worth noting. In a typical silica crucible crystal grown under $H_2$, the rate for both centers is $\sim 10^{-4} \text{ cm}^{-1}$. In the crystal grown directly from a graphite susceptor under $H_2$, the rate is $\sim 2 \times 10^{-5} \text{ cm}^{-1}$. These rates compare to those for donor defects in doped n-type material of $10^{-3} - 2 \times 10^{-2} \text{ cm}^{-1}$\textsuperscript{3}.

These numbers bring into focus results reported by Emtsev et al\textsuperscript{26} on the $\gamma$-irradiation of high-purity germanium, in fact much of the same material used in this work. Using a much higher dose rate ($6 \times 10^{12} \text{ $\gamma$ cm}^{-1} \text{ sec}^{-1}$), and a much higher total dose ($1.5 \times 10^{19} \text{ $\gamma$ cm}^{-2}$), they found two prominent induced defects, acceptors at $E_V + 0.2 \text{ eV}$ and $E_V + 0.084 \text{ eV}$. These levels were ascribed to hydrogen-vacancy complexes, and it was found that the presence of oxygen reduced the rate of formation and the maximum concentration of these defects. Emtsev et al pointed out that both oxygen and hydrogen may form a variety of complex defects, the relative importance of each determined by the ratio of hydrogen and oxygen concentrations and by the thermal conditions during the
crystal growth. The introduction rates of the $E_v + 0.2$ eV and $E_v + 0.084$ eV centers were typically $8 \times 10^{-5} \text{cm}^{-1}$ and $10^{-4} \text{cm}^{-1}$ respectively, which are similar to those observed by us for the oxygen-related $E_v + 0.23$ eV and $E_v + 0.38$ eV centers. Using the DLTS technique which is more sensitive than the Hall effect method used by Emtsev et al, we should have seen these hydrogen-related centers, even at our lower $\gamma$ doses. One can only conclude that the formation of the various oxygen- or hydrogen-related defects is dose rate dependent in high-purity p-type germanium.

CONCLUSIONS

Two deep hole trapping states are induced by low dose rate $\gamma$-irradiation in all p-type high-purity germanium crystals grown for use as nuclear radiation detectors. An extensive study of these centers in crystals grown under widely varying conditions of dopant, crucible material, gas ambient and post-growth annealing leads to the conclusion they are most likely composed of vacancy associations with oxygen. Since their concentrations are always similar (within a factor of three) and they always appear together, the levels may be caused by two charge states of the same defect, but this remains to be verified. The photocapacitance method of Ledebo and Wang\textsuperscript{28} which determines the coupling of defects with different charge states, would be useful in this regard.

It would be interesting to irradiate doped p-type material ($N_A - N_D \sim 10^{15} \text{cm}^{-3}$) at the same dose rate and to the same total dose as that used by Emtsev, et al in their experiments where they postulated that the resulting defects were due to hydrogen-vacancy complexes. Use of doped material would
enable the spectroscopic DLTS method to be used to unambiguously measure the defect levels. One condition for using DLTS is of course that the deep level concentration must be a fraction of the shallow level concentration, which is not a limitation of the Hall effect. Use of the DLTS technique on more highly doped γ-irradiated material would also enable identification of the \( E_v + 0.23 \) eV and \( E_v + 0.38 \) eV centers as acceptors or donors by observation of an associated Poole-Frenkel effect (field enhanced emission of trapped carriers) in p-type material. The high electric field required is not attainable in high-purity material. We have been careful throughout this paper to refer to these centers as hole traps, rather than acceptor defects, as DLTS does not directly give this information.

A further interesting experiment would be the γ-irradiation of germanium-rich Ge-Si alloys. Measurement of the energy level shifts of the two defects as a function of alloy composition would make it possible to determine the extent to which the valence and conduction bands contribute to making up the defect wave functions. Such measurements have proven fruitful in determining the microstructure of defects in ternary mixed crystals. The discovery of a recombination enhanced annealing mechanism in the γ-irradiated high-purity germanium samples may have application to other types of radiation damage in germanium. For example, proton and neutron damaged germanium detectors may be able to be annealed by warming to room temperature and then leaving forward biased for a number of hours.
ACKNOWLEDGMENTS

The authors are indebted to George Gentle, Ron Glover, Gary Burrell and John Gray of the AAEC γ-pond facility, and Dr. Dick Lemon of the LBL γ-irradiation facility for their assistance throughout this experiment. The technical assistance of Alan Williams, Dimitri Alexiev, Peter Lee, and David Garton (AAEC) was much appreciated, while Jim Duffy (Electrical Engineering, UCB) performed the ion-implantations.

This work was supported in part by the Director's Office of Energy Research, Office of Health and Environmental Research, U.S. Department of Energy under Contract No. DE-AC03-76SF00098, and in part by the U.S. National Science Foundation under Contract No. DMR-8203430.
REFERENCES

FIGURE CAPTIONS

Fig. 1. DLTS spectrum from γ-irradiated (dose = 2.8 x 10^{15} \text{ycm}^{-2}) AAEC p-type sample (N_A - N_D = 4 x 10^{10} \text{cm}^{-3} prior to irradiation). Sample reverse bias V_R = 10 V, bias pulse amplitude V_P = 10 V, correlator time constant T_c = 10 ms.

Fig. 2. Arrhenius plots of the two dominant defects observed in γ-irradiated p-type germanium.

Fig. 3. DLTS spectrum from γ-irradiated (dose = 2.8 x 10^{15} \text{ycm}^{-2}) LBL p-type sample (N_A - N_D = 5 x 10^{10} \text{cm}^{-3} prior to irradiation). Spectrum conditions are the same as in Fig. 1.

Fig. 4. DLTS spectrum from γ-irradiated (dose = 2.8 x 10^{15} \text{ycm}^{-2}) GE p-type sample (N_A - N_D = 3 x 10^{10} \text{cm}^{-3} prior to irradiation). Spectrum conditions are same as Figs. 1 and 3.

Fig. 5. DLTS spectrum from copper-doped LBL p-type sample after γ-dose of 3.4 x 10^{15} \text{ycm}^{-2}. Reverse bias V_R = 5 V, correlator time constant T_c = 10 ms. The copper-related defects have not been affected by the γ-irradiation.

Fig. 6. Relative correlator signal output versus pulse width for the two dominant defect levels observed in γ-irradiated p-type germanium. Bias pulse amplitude equals reverse bias in all cases. The slope of the two lines is related to the hole capture cross-section of the two levels at the indicated temperature.

Fig. 7. DLTS spectra from γ-irradiated (dose = 5.6 x 10^{15} \text{ycm}^{-2}) p-type germanium sample (N_A - N_D = 10^{13} \text{cm}^{-3}) grown from an alumina crucible. (i) Lithium-drifted (~ 1 mm) prior to irradiation; and (ii) untouched. Note the reduced density of the γ-induced defects in the presence of lithium. Spectral conditions have been adjusted to investigate the
same depletion depth (~ 80 µm) and give the same relative response.

Fig. 8. (a)(i) DLTS spectrum from γ-irradiated (dose = 3.4 x 10^{15} \text{cm}^{-2}) p-type sample \((N_A - N_D = 2 \times 10^{10} \text{cm}^{-3} \text{ prior to irradiation})\) grown from a silica crucible under H₂. (ii) DLTS spectrum from γ-irradiated (dose = 3.4 x 10^{15} \text{cm}^{-2}) p-type sample \((N_A - N_D = 2 \times 10^{10} \text{cm}^{-3} \text{ prior to irradiation})\) grown from a graphite crucible under H₂. Note the reduced defect density in this material of lower oxygen content.

(b) DLTS spectrum from γ-irradiated (dose = 5 x 10^{16} \text{cm}^{-2}) aluminum-doped \((p = 8 \times 10^{14} \text{cm}^{-3})\) p-type sample grown from a graphite crucible under H₂. No evidence of the two defect states is found in this low oxygen content material.

Fig. 9. DLTS spectra from γ-irradiated (dose = 6.5 x 10^{15} \text{cm}^{-2}) Hoboken crystal grown for lithium drift compensation.

(a) One section showed a low concentration of the hole trapping centers.
(b) Typical spectrum from the other three sections—the defect states were not observed.

Fig. 10. Schematic representation of the 20-minute isochronal annealing in hydrogen ambient of the \(E_V + 0.23 \text{ eV}\) and \(E_V + 0.38 \text{ eV}\) defect states.

Fig. 11. Schematic representation of the 290 K isothermal annealing of the \(E_V + 0.23 \text{ eV}\) and \(E_V + 0.38 \text{ eV}\) defect states under injection current densities of 40, 100 and 200 mA/cm².

Fig. 12. Annealing rate of the \(E_V + 0.23 \text{ eV}\) and \(E_V + 0.38 \text{ eV}\) defect states as a function of injection current density at 290 K.

Fig. 13. DLTS spectra from γ-irradiated (dose = 2.8 x 10^{15} \text{cm}^{-2}) GE p-type sample showing sensitivity of the \(E_V + 0.23 \text{ eV}\) center to infra-red illumination from a GaAs light emitting diode. (i) Sample cooled to 77 K, no infra-red illumination; and (ii) sample heated to ambient with infra-red illumination, spectra are recorded with all conditions identical.
Table I.

High-purity p-type germanium crystals used in these experiments. All crystals were grown by the Czochralski method unless otherwise indicated.

<table>
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<tr>
<th>Supplier</th>
<th>No. of Crystals</th>
<th>( N_A-N_D ) (cm(^{-3}) at 77K)</th>
<th>Gas Atmosphere</th>
<th>Crucible Materials</th>
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<tr>
<td>LBL</td>
<td>15</td>
<td>( 1 \times 10^{10} - 7 \times 10^{11} )</td>
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<td>( \text{H}_2 )</td>
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<td>( 8 \times 10^{15} - 1 \times 10^{16} ) (Zn doped)</td>
<td>unknown</td>
<td>zone levelled</td>
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Fig. 1.

**AAEC 78-93-4**

$V_R = 10V, V_P = 10V(200\mu S)$

$\tau_C = 10mS$

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<th>Sample Temperature (K)</th>
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$E_V + 35$ meV
$E_V + 87$ meV
$E_V + 69$ meV
$E_V + 0.10$ eV
$E_V + 0.23$ eV
$E_V + 0.38$ eV

GAIN X2
Fig. 2.
Fig. 3.
Fig. 4.

GE Co #7
$V_R=10V, V_P=10V(200\mu S)$
$\tau_C=10mS$
CORRELATOR OUTPUT
(ARBITRARY UNITS)

SAMPLE TEMPERATURE (K)

CRBL 438-2.3
V_r=10V, V_p=10V(200\mu S)

T_c=10mS

E_v + 0.23eV

E_v + 0.33eV

E_v + 0.38eV

Fig. 5.
Fig. 6.

- $E_v + 0.38\text{eV}$
  $T = 163\text{K}$

- $E_v + 0.23\text{eV}$
  $T = 103\text{K}$
80-23-4 Al₂O₃ CRUCIBLE
Vₐ=30V, Vₚ=30V(200µS)
Tₖ=10mS
W=80µm

Eᵥ+0.23eV
Nₜ=2.5%(Nₐ-N₅)

Eᵥ+0.38eV
Nₜ=0.15%(Nₐ-N₅)

(i) Li DRIFTED,
(ii) UNTouched

Fig. 7.
(a) LBL Ge (i) 606-6.6
(ii) 564-6.6
$\tau_c = 10 \text{mS}$

(b) AAEC Ge 79-9-2
Al DOPED (10 cm)
$\tau_c = 10 \text{mS}$

Fig. 8.
(i) HOBOKEN 'R' #1
\[ V_R = 15V, \ V_P = 15V(200\mu S) \]
\[ \tau_C = 10mS \]

\[ N_T - 3 \times 10^{-4}(N_A - N_D) \]

(ii) HOBOKEN 'R' #2
\[ V_R = 15V, \ V_P = 15V(200\mu S) \]
\[ \tau_C = 10mS \]

\[ N_T < 10^{-4}(N_A - N_D) \]

SAMPLE TEMPERATURE ([K])

CORRELATOR OUTPUT (ARBITRARY UNITS)

Fig. 9.
HP P-TYPE $\gamma$-IRRADIATED Ge

- $E_v + 0.38eV$
- $E_v + 0.23eV$

Fig. 10.
$E_V + 0.23\text{eV, } E_V + 0.38\text{eV}$

290K

$\triangle 40\text{mA/cm}^2$

$\times 100\text{mA/cm}^2$

$\bullet 200\text{mA/cm}^2$

Fig. 11.
Fig. 12.

\[ E_V + 0.38 \text{eV}, E_V + 0.23 \text{eV} \]

290K

INJECTION CURRENT DENSITY (mA/cm\(^2\))

ANNEALING RATE (MIN\(^{-1}\))
Fig. 13.
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