PHYSICAL REVIEW VOLUME 174, NUMBER 3 15 OCTOBER 1968

Energy and Orientation Dependence of Electron-Irradiation-Induced Damage in Undoped GaSb*

K. Thommen
Atomics International, A Division of North American Rockwell Corporation, Canoga Park, California
(Received 17 June 1968)

Undoped GaSb single crystals were irradiated at 77°K with electrons having energies betwen 0.23 and 1.00 MeV. The incident electron beam was aligned with either the [111] or the [$\bar{1}\bar{1}\bar{1}$] direction of the sample. The isochronal recovery of the Hall coefficient and the electrical resistivity was measured after each irradiation. The positions and widths of the four major recovery stages were independent of the irradiation energy. The production rates of the defects corresponding to the four stages differed in their energy dependence. Their threshold electron energies were evaluated as 0.34 MeV for stage I, 0.17 MeV for stage II, 0.33 MeV for stage III, and 0.34 MeV for stage IV. Stage-I defect production was higher with the electron beam in the [111] direction than in the [$\bar{1}\bar{1}\bar{1}$] direction. The reverse was the case for stage-II, -III, and -IV defect production. Stage I is attributed to the recovery of defects created by the displacement of single Sb atoms. The threshold displacement energy for Sb atoms is larger than 7.5 eV but smaller than 13.1 eV. Stage II is interpreted as recovery of defects produced by the displacement of single Ga atoms. The threshold displacement energy for Ga atoms is about 6.2 eV. It is suggested that stages III and IV represent the recovery of defects created by multiple displacement processes starting with the displacement of a Ga atom.

I. INTRODUCTION

PREVIOUS investigation[1] of the effect of 1-MeV electron irradiation on the electrical properties of undoped GaSb has shown that most of the damage recovers upon isochronal pulse annealing in four distinct stages. Their center temperatures were 122, 163, 203, and 365°K, and they were labeled as stages I–IV in order of increasing temperature. No attempt has been made to assign specific defect configurations to these stages. In the present work, the amount of recovery of the reciprocal Hall mobility occurring in each of these stages was studied for various energies of the bombarding electrons and for two orientations of the monocrystalline samples with respect to the incident electron beam. There is no doubt that irradiation with 1-MeV electrons will cause displacement of both Ga and Sb atoms. However, by lowering the electron energy gradually, one should eventually enter an energy range where only Ga atoms can be displaced. This can be expected because a Ga atom is lighter than an Sb atom by a factor of 1.7, and the maximum amount of energy which can be transferred from an electron to an atom in an elastic collision is inversely proportional to the atomic mass. Thus, by studying the energy dependence of the recovery spectrum one may obtain information about which stage or stages correspond to defect configurations created by the displacement of Ga atoms. However, if there are various types of Ga defects having widely different threshold displacement energies, only those with the lowest threshold energy can be identified in this way.

The threshold electron energy for the production of the defects that recover in a given stage can be determined by measuring the amount of recovery in this stage per unit dose as a function of the irradiation energy. Provided that it is known which type of atom is displaced in producing these defects and that the displaced atom is the same as the one which was struck by the incident electron, the threshold displacement energy for this atom can be evaluated. The results of the present work allow determination of the threshold electron energies for the four stages. The threshold displacement energy for Ga atoms is evaluated, but for Sb atoms it is not clear if the latter of the two conditions stated above is fulfilled. Briefly, this uncertainty exists because the threshold electron energies for the direct and indirect displacement of Sb atoms can be expected to be not too different. In the indirect process, the electron transfers energy to a Ga atom which subsequently strikes and displaces a neighboring Sb atom. This subject will be discussed in more detail in Sec. IV.

Eisen,[2] who studied the electron-irradiation-induced damage in InSb, first realized that measurement of the orientation dependence of the damage production in III-V compounds can be a useful tool for distinguishing between defects produced by the displacement of group-III atoms and group-V atoms. Most of the III-V compound semiconductors, including GaSb, crystallize in the zinc-blende structure for which the ⟨111⟩ axis is polar. A portion of a model of the zinc-blende lattice is shown in Fig. 1. The black spheres represent group -III atoms and the white ones represent group-V atoms. The atoms 1, 2, 15, and 16 are along a ⟨111⟩ axis. Distinction is made between the [111] and [$\bar{1}\bar{1}\bar{1}$] directions as indicated.[3] It seems plausible to assume that it is easier to displace atom 2, which happens to be a group-III atom, in the [$\bar{1}\bar{1}\bar{1}$] direction than in the [111] direction for the following reasons: If its initial momentum coincides with the [111] direction, it will immediately undergo a collision with atom 1, lose most of its energy, and most likely relax

* Based on work supported by the Metallurgy Branch, Division of Research, U. S. Atomic Energy Commission, under Contract No. AT(04-3)-701.

[1] K. Thommen, Phys. Rev. **161**, 769 (1967).

[2] F. H. Eisen, Phys. Rev. **135**, A1394 (1964).

[3] The same notation as in Ref. 2 is used in this paper.

back into its original position. On the other hand, if its initial momentum coincides with the [$\bar{1}\bar{1}\bar{1}$] direction, it may squeeze through the triangular lens formed by the atoms 3-4-5 and possibly through the lenses 6-7-8 and 9-10-11, and only then collide with atom 15. There are two interstices which, from a purely geometrical point of view, could conveniently accommodate the displaced atom. They are in the center of the tetrahedra 2-6-7-8 and 9-10-11-15, respectively. Although it is doubtful that the former interstice would provide a stable position for the displaced atom, the second one may very well do so. If the initial momentum of atom 2 is directed at a small angle with either the [111] or [$\bar{1}\bar{1}\bar{1}$] direction, the above considerations still should hold. Likewise, it should be easier to displace a group-V atom near the [111] direction than near the [$\bar{1}\bar{1}\bar{1}$] direction. Suppose a single crystal specimen to be irradiated with a collimated electron beam in the [111] direction. Eisen[2] has shown that then for an electron energy not too far above the threshold energy for group-V atom displacements, one can expect the production rate of defects created by the displacement of group-V atoms to be higher than that measured with the electron beam aligned in the [$\bar{1}\bar{1}\bar{1}$] direction. On the other hand, for an electron energy not too far above the threshold energy for group-III atom displacements, the production of defects created by the displacement of group-III atoms should proceed at a higher rate for irradiation in the [$\bar{1}\bar{1}\bar{1}$] direction than for irradiation in the [111] direction.

In the present work, GaSb single crystals were irradiated in the [111] and [$\bar{1}\bar{1}\bar{1}$] directions, and the amount of recovery per unit dose occurring in each of the four stages was measured. Orientation effects were observed for all stages.

II. EXPERIMENTAL PROCEDURES

Single-crystal wafers of undoped GaSb polished to a Linde B finish were purchased from Bell & Howell, Pasadena, California. Their thickness was between 0.012 and 0.017 cm, and the surface normals of the large areas were close to one of the ⟨111⟩ axes. The material was p type, and its electrical characteristics were the same as in earlier work of this author.[1] Bridge-shaped samples were cut using a sand-blasting technique. They were mounted on a copper base plate in the same manner as described in Ref. 1. The orientation of the

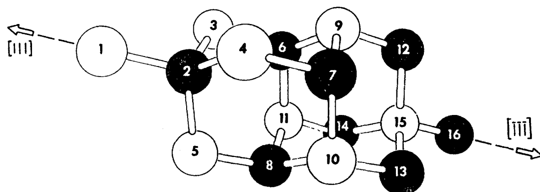

FIG. 1. Portion of a model of the zinc-blende structure. The numbers 1–16 label atom positions referred to in the text.

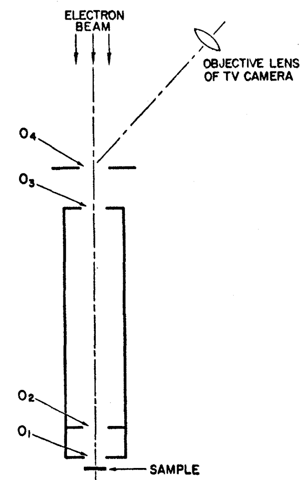

FIG. 2. Apertures for electron-beam collimation (schematic).

mounted samples with respect to the base plate was checked with x rays. Distinction between the Sb (111) surface and the Ga ($\bar{1}\bar{1}\bar{1}$) surface was achieved with a preferential-etching technique described by Faust and Sagar.[4] This etching was performed after completion of the irradiation experiments.

Since previous recovery studies[1] have shown that no observable recovery of the radiation-induced damage occurred during annealing from 15 to 110°K, in the present study all irradiations and measurements were made at 77°K. A liquid-nitrogen cryostat was used which was similar to the one described by Eisen,[5] except for the beam-defining apertures. In this work collimation of the electron beam was effected by a set of three rectangular apertures shown schematically in Fig. 2 as O_1 to O_3. For irradiations with electrons having energies of 0.5 MeV or more, O_3 was covered with a 6-μ-thick aluminum scattering foil. The distance between O_3 and the sample was about 50 cm. For energies below 0.5 MeV, no scattering foil was used, but a circular aperture O_4 was placed above O_3. The upper side of O_4 was coated with luminescent paint so that the position of the electron beam at O_4 could be monitored through a glass window with a closed-circuit television camera. The beam position could then be adjusted by two pairs of steering coils to give uniform illumination of O_3. The electron beam was produced by a Van de Graaff generator. Energy selection was achieved by a double 60° magnet system. The irradiated sample was annealed isochronally *in situ* from 77 to 420°K. Temperature increments and annealing times were chosen such that the isochronal annealing approximately corresponded to an annealing with a constant warm-up rate of 1°K/min.

For the study of the energy dependence of the radiation-induced damage, a sample was irradiated at a particular energy, subsequently isochronally annealed, then irradiated at another energy, and so on. For the

[4] J. W. Faust and A. Sagar, J. Appl. Phys. **31**, 331 (1960).
[5] F. H. Eisen, Phys. Rev. **123**, 736 (1961).

study of the orientation dependence of the damage, the sample was first irradiated and annealed at various energies in one position, for instance, with the (111) surface exposed to the electron beam. Then the sample was inverted together with the copper base plate, so that now the $(\bar{1}\bar{1}\bar{1})$ surface was exposed to the electron beam. The same irradiations and anneals were then performed as with the sample in the first position. Resistivity and Hall measurements were made at 77°K using standard potentiometric techniques. The magnetic field was 530 G.

From these measurements, an apparent carrier concentration $p=1/eR_H$ and the Hall mobility $\mu_H=R_H/\rho$ were evaluated. Here, ρ is the resistivity, R_H is the Hall coefficient, and e is the absolute value of the electronic charge. Let $(\Delta\mu_H^{-1})_i$ denote that portion of the radiation-induced change of the reciprocal Hall mobility which recovered in the ith recovery stage. $(\Delta\mu_H^{-1})_i$ was chosen as a measure of the concentration of those defects which were removed in the ith recovery stage. The change of the reciprocal Hall mobility was found to be a more suitable measure for defect concentrations than the corresponding changes of the carrier concentration or the resistivity, because it showed a higher defect sensitivity and could be measured with higher reproducibility than those. The latter is true because the Hall mobility had a considerably weaker temperature dependence than the resistivity and the carrier concentration. Therefore, it was less sensitive to small changes of the temperature at which the resistivity and the Hall voltage were measured. Changes of this temperature of the order of a few hundredths of a degree occurred from one day to another. The reproducibility of $(\Delta\mu_H^{-1})_i$ was 3×10^{-7} V sec cm^{-2}. This corresponds to an approximate defect concentration of 2×10^{14} cm^{-3}. For most irradiations, exposures were chosen such that the total change of the reciprocal Hall mobility $\Delta\mu_H^{-1}$ was about 7×10^{-5} V sec cm^{-2}. Only for energies $\leqq0.3$ MeV were the values of $\Delta\mu_H^{-1}$ smaller. For some irradiations near the threshold energies of stages I, III, and IV, doses were applied which resulted in values for $\Delta\mu_H^{-1}$ up to 3×10^{-4} V sec cm^{-2}.

For irradiations with 1.0-MeV electrons, it was ascertained that $(\Delta\mu_H^{-1})_i$ varied linearly with irradiation dose up to a dose resulting in $\Delta\mu_H^{-1}=3\times10^{-4}$ V sec cm^{-2}. A linear variation of $(\Delta\mu_H^{-1})_2$ with irradiation dose up to $(\Delta\mu_H^{-1})_2=3\times10^{-5}$ V sec cm^{-2} was observed for irradiation with 0.3-MeV electrons. At this energy, all damage recovered in stage II. It is believed that a linear relationship between $(\Delta\mu_H^{-1})_i$ and irradiation dose also exists for other irradiation energies.

III. RESULTS

A. Energy Dependence

In this section, we present data on the energy dependence of the defect production rates η_i, $i=1$–4,

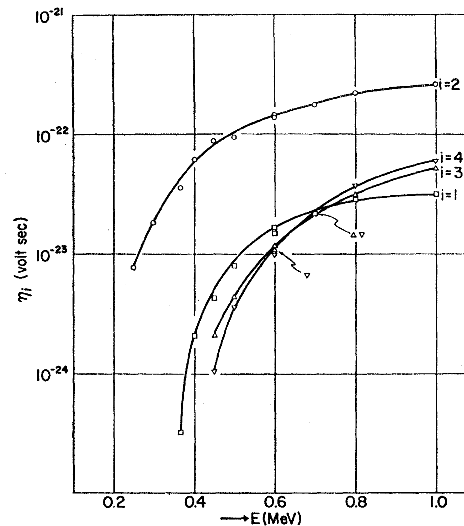

FIG. 3. Production rates η_i, $i=1$–4, versus electron energy E.

the center temperatures T_{ci} and widths ΔT_i of the four recovery stages, and the ratio $(\Delta\mu_H^{-1})_2/\Delta p_2$. η_i is defined as $(\Delta\mu_H^{-1})_i/\phi$, where $(\Delta\mu_H^{-1})_i$ has the meaning as defined in Sec. II, and ϕ is the irradiation dose. T_{ci} is the temperature at which 50% of the recovery of the reciprocal Hall mobility in the ith stage has been completed, ΔT_i is the difference of the temperatures corresponding to the 90 and 10% completion of the recovery in the ith stage, and Δp_i is the recovery of the carrier concentration in the ith stage. T_{ci}, ΔT_i, and $(\Delta\mu_H^{-1})_i/\Delta p_i$ are considered to be quantities which can be used to distinguish between various types of defects. η_i measures the production efficiency of a given type of defect.

All irradiations reported in this section were performed with sample 14, which was mounted so that the [111] axis formed an angle of 10° with the beam direction. The electron energy was varied between 0.25 and 1.00 MeV.

Figure 3 shows the defect production rates η_i as functions of the electron energy E. Stage II has the highest production rate throughout the entire energy range. With decreasing energy, the stage-II defects represent an increasing fraction of the total number of defects. At 0.25 and 0.30 MeV only stage-II defects could be detected. Stage-I defects apparently have a higher threshold electron energy, but otherwise their production rate exhibits an energy dependence similar to that of the stage-II defect production rate. Within the energy range 0.6–1.0 MeV, the ratio of η_1/η_2 is nearly constant. η_3 and η_4 have threshold energies close to that of η_1. However, they show less saturation tendency at higher energies than η_1 and η_2. Consequently, stage-III and stage-IV defects increase in relative importance as the irradiation energy increases.

Center temperatures T_{ci} and widths ΔT_i of the four annealing stages are shown for various irradiation

TABLE I. Center temperatures T_c and widths ΔT of recovery stages I–IV, in °K, and ratio $(\Delta\mu_H^{-1})_2/\Delta p_2$ for various irradiation energies E.

Stage	E (MeV)	0.25	0.30	0.37	0.40	0.45	0.50	0.70	1.0
I	T_c	...	...	123±2	...	124±1	124±1	124±1	124±1
	ΔT	...	...	11±4	...	11±2	12±2	11±2	11±2
II	T_c	169±1	170±1	...	170±1	170±1	170±1	169±1	170±1
	ΔT	15±2	16±1	...	15±1	16±1	15±1	15±1	15±1
	$(\Delta\mu_H^{-1})_2/\Delta p_2 \times 10^{-21}$ (V sec cm)	52±10	45±5	43±4	48±5	46±4	44±4	44±4	50±5
III	T_c	...	...	...	...	216±3	214±2	212±1	212±1
	ΔT	...	...	...	...	19±7	24±6	22±4	23±2
IV	T_c	...	...	...	...	357±4	355±3	360±2	357±1
	ΔT	...	...	...	...	29±7	29±5	33±5	38±4

energies in Table I.[6] It can be seen that for each stage T_{ci} and ΔT_i are energy-independent within the error limits. The widths of stages III and IV have a particularly large error, which is due to the small size of these stages. Therefore, the data do not exclude a slight narrowing of stages III and IV with decreasing energy.

Also shown in Table I are values of $(\Delta\mu_H^{-1})_2/\Delta p_2$ for various energies. They indicate that this quantity is independent of the irradiation energy within the error limits. The nonsystematic fluctuations are probably due to the inaccuracy in the measurements of Δp_2. For the other stages $(\Delta\mu_H^{-1})_i/\Delta p_i$, $i=1, 3, 4$, could not be determined with reasonable accuracy because of the small size of Δp_i and the correspondingly large errors.

B. Orientation Dependence

The defect production rates η_i, $i=1$–4, were measured at various irradiation energies ≤ 0.5 MeV for two sample orientations. In one orientation the direction of the incident electron beam coincided nearly with the [111] direction of the sample, in the other nearly with the $[\bar{1}\bar{1}\bar{1}]$ direction. Most data were taken with sample 16 for which the misalignment was 2°. However, a few measurements were made with sample 14, which showed a misalignment of 10°. The defect production rates corresponding to the two orientations will be denoted η_i^+ and η_i^-. We define the anisotropy A_i by

$$A_i = (\eta_i^+ - \eta_i^-)/(\eta_i^+ + \eta_i^-).$$

Figure 4 shows A_i for various energies. Although the observed anisotropy effects are quite small, it is evident that the sign of the stage-I anisotropy is positive and the sign of the anisotropy of the other three stages is negative. This indicates that stage-I defects are more easily created with the electron beam in the [111] direction, whereas stage-II–IV defects are more easily produced in the $[\bar{1}\bar{1}\bar{1}]$ direction. For stages I and II the data of Fig. 4 indicate a decrease of the anisotropy with increasing irradiation energy. For the stage-III and -IV anisotropies, insufficient data are available to draw conclusions about their energy dependence.

C. Threshold Energies

Plots of the defect production rates η_i versus the electron energy E on linear scales would exhibit strong positive curvatures at low energies for all four stages. Therefore, determination of the threshold energies by extrapolation of these curves would be very inaccurate. Bäuerlein[7] argued that, for a "thick" sample within a limited energy range above threshold, the square root of the production rate for a given type of defect should vary linearly with $E-E_{\text{th}}$, where E_{th} is the threshold electron energy. Although his arguments were based on considerable simplifications, a square-root dependence of the defect production rate was indeed observed in several cases, and threshold electron energies were evaluated by linear extrapolation.[2,7,8]

Figure 5 shows square-root plots of the production

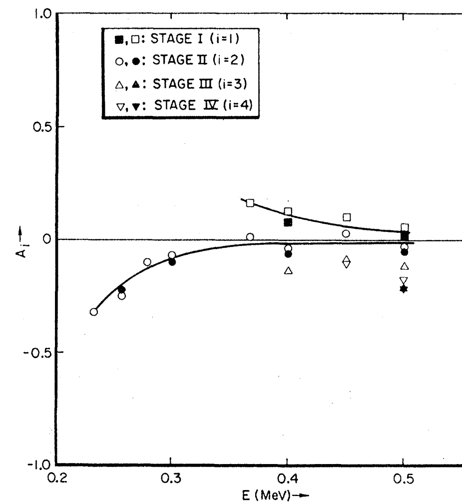

FIG. 4. Anisotropies A_i, $i=1$–4, versus electron energy E. Full symbols: sample 14; open symbols: sample 16.

[6] The center temperatures reported in this paper are slightly different from those quoted in Ref. 1. For stages I, II, and III this difference can be accounted for quantitatively by the faster average warmup rate used in the present work. In the temperature region of stage IV, however, the annealing procedures were approximately the same and the reason for the different position of this stage remains unclear.

[7] R. Bäuerlein, *Radiation Damage in Solids* (Academic Press Inc., New York, 1962), p. 358; Z. Physik **176**, 498 (1963).

[8] F. J. Bryant and E. Webster, Phys. Status Solidi **21**, 315 (1967).

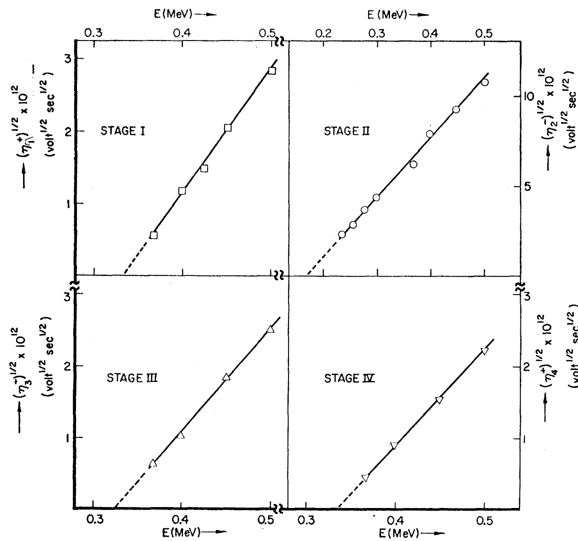

FIG. 5. Square root of production rates $(\eta_i^+)^{1/2}$ or $(\eta_i^-)^{1/2}$, $i=1$–4, versus electron energy E.

rates for the four stages of GaSb. These data were obtained with sample 16. Since we wish to evaluate the threshold electron energies, the production rates for irradiation in the "easy" direction were plotted. In all four cases the data can be well fitted by straight lines. The threshold electron energies as determined by linear extrapolation are stage I, 0.34 MeV; stage II, 0.17 MeV; stage III, 0.33 MeV; and stage IV, 0.34 MeV.

IV. DISCUSSION

The main purpose of this work is to assess which portions of the electron-irradiation-produced damage in GaSb is related to the displacement of Ga and Sb atoms, respectively, and to determine the minimum amount of energy which has to be imparted to either of these two atoms to effect its permanent displacement. In this context, we are interested only in those defects which were created as the result of a collision process between a bombarding electron and a regular lattice atom. Therefore, it is of great importance to ascertain that the defects, which we observe after electron irradiation, are indeed created by such collision processes and not by some other defect-producing mechanism. Our results indicate that defects can be produced by electron irradiation with an energy as low as 0.17 MeV. If this value is interpreted as the minimum electron energy necessary to displace a Ga atom, the corresponding threshold displacement energy is 6.2 eV, which is less than half of the lowest values reported for the threshold displacement energies in silicon and germanium. One cannot help but suspect that some defects might be produced by a subthreshold mechanism and not by collision-induced displacement of regular lattice atoms. The two most likely subthreshold mechanisms are the following:

(a) Displacement of a regular lattice atom as the result of a radiochemical process initiated by the ionizing action of the bombarding electrons. It has been suggested[9,10] that a mechanism of the type proposed by Varley[11] for F-center production in alkali halides may be operative in III-V compounds.

(b) Imperfection-assisted defect production. This could be collision-induced or ionization-induced displacement of either an impurity atom or a host lattice atom in the vicinity of a lattice imperfection.

In the following paragraphs we shall discuss in some detail the possibility that some of the damage observed after electron irradiation may be produced by either of the two above-mentioned mechanisms. It is important for this discussion to recall that for each of the four stages the center temperature T_{ci} and the width ΔT_i were independent of the irradiation energy. For stage II $(\Delta\mu_H^{-1})_2/\Delta p_2$ was also found to be energy-independent. This is considered as evidence that for a given stage the same class of defects was monitored throughout the entire investigated energy range.

Only electrons with initial energies lower than 0.30 MeV are stopped in a 0.013-mm-thick sample. For higher irradiation energies, the amount of energy which is dissipated in the sample per unit dose in the form of ionization and electronic excitation decreases slightly with increasing energy of the incident electrons. Therefore, one can expect that, at least for irradiation energies larger than 0.30 MeV, the production rate for defects which are created by an ionization-based mechanism would decrease slightly with increasing energy.[12,13] However, in the present work an increase of the defect production rate of one to two orders of magnitude was observed with increasing energy for each of the four stages. This proves that none of the four stages can be associated with the recovery of defects which are produced entirely or predominantly by ionization. It remains to examine the possibility of a small background defect production for a given stage caused by ionization. As far as stages I, III, and IV are concerned, irradiations were performed at energies lower than the indicated threshold electron energies, and such a background would have been detected if it were larger than 30% of the lowest production rate measured above threshold. Unfortunately, such a test was not possible for stage II because the accelerator could not be operated at energies lower than 0.23 MeV. Therefore, an estimate of the upper limit of a possible background contribution must be based on examination of the

[9] G. W. Arnold and F. L. Vook, Phys. Rev. **137**, A1839 (1965).
[10] R. Yu. Khansevarov *et al.*, Phys. Status Solidi **22**, K95 (1967).
[11] J. H. O. Varley, Nature **174**, 886 (1954); J. Nucl. Energy **1**, 130 (1954); J. Phys. Chem. Solids **23**, 985 (1962).
[12] For alkali halides where defect production is dominated by ionization-based processes, it has been shown (Ref. 13) that the number of F centers produced per unit absorbed energy decreases by about 30% from ~50 keV (filtered x rays) to 2.0 MeV (electrons).
[13] V. H. Ritz, Phys. Rev. **133**, A1452 (1964).

energy dependence of the production rate. Although there is no rigorous proof that for a collision-based defect production mechanism the square root of the production rate should vary linearly with energy, we shall assume that this is the case. Addition of a background defect production rate which is either energy-independent or decreases with increasing energy would then cause deviation from straight-line behavior in a $\eta_2^{1/2}$-versus-E plot at low energies. Examination of the data of Fig. 5 shows that deviation from straight-line behavior would be noticeable if 50% or more of the production rate at 0.23 MeV were due to a background effect. The maximum error in the threshold energy for stage II, resulting from a background contribution, is then of the order of 10%. Other evidence that the stage-II production rate has no significant background comes from the orientation measurements. Ionization-based defect production should be orientation-independent. If it would contribute significantly to the stage-II defect production at about 0.25 MeV, A_2 would decrease with decreasing energy in this region. However, the stage-II anisotropy is largest at the lowest energies. Likewise, the energy dependence of A_1 indicates that ionization-induced defect production is negligible for stage-I defects.

Next we consider the case that either an impurity atom or a host lattice atom near some lattice imperfection becomes displaced by a collision mechanism. If an impurity atom is much lighter than a Ga atom, a correspondingly larger amount of energy can be transferred to it by the bombarding electron. The impurity atom can either become a displaced atom itself or it can transfer a portion of its energy to a neighboring host lattice atom and effect its displacement. The threshold electron energy for either process can be lower than that for the direct displacement of a host lattice atom. On the other hand, atoms in the vicinity of lattice irregularities, such as impurity atoms, stoichiometry defects (Ga atoms on Sb sites), dislocations, or grain boundaries, may be more weakly bound to their lattice positions and, consequently, may require less energy for permanent displacement. The total concentration of such "atypical" atoms, including impurities, can be expected to be not larger than 10^{18} cm^{-3}. From the radiation-induced change of the Hall mobility it is estimated that the production rate for stage-II defects at 1.0 MeV is of the order of 1 cm^{-1}. Application of simple displacement theory, assuming an isotropic threshold displacement energy of $T_{0,\mathrm{Ga}}=6.2$ eV, yields a value of 7 cm^{-1} for the production rate of Ga Frenkel defects at 1.0-MeV electron energy if multiple displacements are neglected. Inclusion of multiple displacements would increase this figure by not more than 30%. Thus, even if we assume that the vacancy and the interstitial of each Frenkel defect act as two independent scattering centers, the measured damage rate for stage-II defects at 1 MeV is only smaller by one order of magnitude than the calculated production rate for Ga Frenkel defects. However, it is quite certain that the displacement energy is not isotropic as assumed in the calculation. Since the experimentally determined value of 6.2 eV is to be interpreted as the minimum displacement energy, the calculated value for the defect production rate is too high. Overestimation by a factor of 5–10 does not seem unreasonable. For instance, with an "effective" displacement energy of 17 eV the production rate for Ga Frenkel defects becomes 1.6 cm^{-1}.

On the other hand, assuming a concentration of atypical atoms of 1×10^{18} cm^{-3} and an isotropic displacement energy not smaller than 6.2 eV, we find that 1.0-MeV electrons produce defects by means of collision-induced displacement of atypical atoms at a rate $\leq 4\times10^{-4}$ cm^{-1}.[14,15] Note that this upper limit is more than three orders of magnitude smaller than the stage-II defect production rate estimated from the observed Hall-mobility change. This discrepancy is not significantly reduced at lower electron energies. For an irradiation energy of 0.25 MeV it still amounts to three orders of magnitude. We conclude, therefore, that stage II cannot be due to the recovery of defects produced by the collision-induced displacement of atypical atoms. Based on analogous arguments, we extend this conclusion to the other three stages.

Finally, we have to consider the case that some defects might be related to the displacement of atypical atoms by an ionization-based mechanism. The possibility that such an effect contributes significantly to the defect production by electron irradiation can be ruled out by the same arguments which were used in discussing the displacement of regular atoms by an ionization-based mechanism.

We have thus come to the conclusion that neither an ionization-based mechanism nor an imperfection-assisted mechanism can account for the production of the stage-I to stage-IV defects at low energies. Therefore, we feel justified in assuming that each of the four types of defects is produced by a process involving the collision of a bombarding electron with a regular lattice atom. Note that this does not exclude the possibility that the displacement process involves the simultaneous occurrence of ionization and collision-induced energy transfer. Since none of the four stages agree in both threshold electron energy and sign of anisotropy, we conclude that the defects corresponding to one stage are produced independently from the defects of any

[14] It should be noted that energy transport over long distances by means of focused collision sequences is unlikely in a material with zinc-blende structure. Therefore the possibility can be excluded that initially energy is transferred from a bombarding electron to a regular lattice atom and subsequently transported to a distant "atypical" atom by a focused collision sequence. Such a mechanism was invoked to explain subthreshold damage in copper (see Ref. 15).

[15] W. Bauer and A. Sosin, J. Appl. Phys. 35, 703 (1964).

other stage.[16] This is substantiated by the observation that at 1 MeV the fractional recovery in a given stage is independent of the irradiation dose. In a previous paper,[1] this author has conjectured that, during the annealing from 77 to 218°K some of the radiation-produced defects interact with defects being present in the material before irradiation, and that the trapped defects are released during the annealing from 340 to 440°K. This interpretation is not consistent with the conclusion in the present paper that the four types of defects, corresponding to the four recovery stages, are produced independently of each other. The conjecture made in Ref. 1 was based on an analysis of the temperature dependence of the Hall coefficient after 1-MeV electron irradiation and subsequent annealing at 218°K, assuming that only one radiation-produced acceptor level influences the carrier concentration. It is possible that this assumption is a too drastic simplification, and therefore it may lead to an erroneous result. It seems to us that the conclusion in the present paper, concerning the independent production of the four types of defects, rests on firmer ground than the conjecture made in Ref. 1. Therefore, we prefer to renounce the latter.

The orientation-dependence measurements yielded the result that, at least in the vicinity of the threshold energy, the production rate of stage-I defects is larger for irradiation in the [111] direction than for irradiation in the $[\bar{1}\bar{1}\bar{1}]$ direction. The opposite behavior was found for the other three stages. Based on the same reasoning which Eisen[2] presented for the case of InSb, we conclude that the stage-I defects are related to the displacement of Sb atoms, whereas the stage-II–IV defects involve displacement of Ga atoms. The conclusion that stage II can be attributed to defects created by the displacement of Ga atoms is consistent with the observation that stage II has the lowest threshold electron energy. Because of the large mass difference between Ga and Sb atoms, one can expect the displacement of Ga atoms to start at substantially lower electron energies than the displacement of Sb atoms, provided that the Ga threshold displacement energy is not larger than the Sb threshold displacement energy. It will be shown below that the threshold displacement energy for Ga is in fact somewhat lower than that for Sb.

For stages I and II, sufficient data are available to show that the magnitude of the anisotropy for these stages decreases with increasing energy and has practically vanished at 0.5 MeV. This can be interpreted as being due to the effect of multiple scattering of the bombarding electrons while they penetrate the sample. Multiple scattering causes a progressive dispersion of the electron beam and therefore tends to conceal the orientation dependence of the defect production rates. At electron energies well above the threshold energy of a given type of defect, these defects will be produced throughout the entire sample. However, the original beam orientation will be wiped out within a penetration depth of the order of 0.003 cm in GaSb. Therefore, the majority of the defects will be produced by electrons with a broad directional distribution, and the observable anisotropy of the production rate will be low. On the other hand, at energies of the incident electrons close to threshold energy defects are produced only in a thin layer near the exposed surface, since the electrons, in addition to being scattered, lose energy as they penetrate the sample. The closer the energy of the incident electron is to the threshold energy the better will the original beam orientation be retained within that portion of the sample in which defects are produced. Provided the displacement energy has an absolute minimum in the "easy direction," i.e., $[\bar{1}\bar{1}\bar{1}]$ for Ga atoms and [111] for Sb atoms, the observable anisotropy of the defect production rate will approach its maximum value of unity as the electron energy approaches the threshold energy. While we can thus explain qualitatively the observed energy dependence of the stage-I and stage-II anisotropies, we cannot account quantitatively for their magnitudes. This would require information about the magnitude of the Ga and Sb displacement energies not only for directions near the $\langle 111 \rangle$ axis but also in all other directions. Such information is not available at the present time.

The threshold electron energy for the production of stage-I defects was found as 0.34 MeV. However, evaluation of the Sb threshold displacement energy $T_{0,\mathrm{Sb}}$ poses some problems since two competing displacement processes have to be considered, both of which are qualitatively consistent with the observed orientation dependence.

(1) The incident electron transfers energy directly to the Sb atom. If the transferred energy is larger than $T_{0,\mathrm{Sb}}$ and the direction of the transferred momentum is close to the [111] direction, the Sb atom will be permanently displaced.

(2) The incident electron transfers energy to a Ga atom which has an Sb atom as nearest neighbor in the [111] direction. If the direction of the momentum transferred to the Ga atom is reasonably close to the [111] direction, a fraction of the energy which was imparted to the Ga atom will be transferred to the neighboring Sb atom. There will be a focusing effect causing the direction of the momentum of the Sb atom to be closer to the [111] direction than the momentum of the Ga atom. If the energy transferred to the Sb atom is larger than $T'_{0,\mathrm{Sb}}$, the Sb atom will be displaced,

[16] The values for the threshold electron energies of stages III and IV are very close, and one may wonder if they are really different. We believe that this is indeed so because, for irradiations in either the [111] or the $[\bar{1}\bar{1}\bar{1}]$ direction, the production-rate ratio η_3/η_4 increases quite drastically with decreasing irradiation energy below 0.5 MeV.

while the Ga atom probably will relax into a place close to its normal lattice position.[17]

We shall refer to these two processes as the direct and indirect displacement process, respectively. If the observed threshold energy for stage-I defects could be attributed to the onset of the direct process, an Sb displacement energy of 8.0 eV would be indicated. On the other hand, if it is the indirect process which starts at the stage-I threshold electron energy, we would have to conclude that the electron must transfer an energy of 14.3 eV to the Ga atom in order to achieve permanent displacement of the Sb atom. How much of this energy is transferred to the Sb atom in a head-on collision depends on the details of the interaction forces between the Ga atom and *all* other atoms in its vicinity. Since these details are not sufficiently understood at the present time, only an upper and a lower limit for the energy transferred to the Sb atom can be established. An upper limit is obtained by assuming that both the Ga and the Sb atom behave like free particles. In this case, if the energy of the Ga atom is 14.3 eV, the amount of energy transferred to the Sb atom would be 13.1 eV. In order to establish a lower limit for $T'_{0,\mathrm{Sb}}$ we assume that the Ga atom loses an energy equal to $T_{0,\mathrm{Ga}}$ because of its interaction with neighboring atoms other than the knock-on Sb atom. This leads to a lower limit of 7.5 eV for $T'_{0,\mathrm{Sb}}$. Clearly the true value of $T'_{0,\mathrm{Sb}}$ will be neither close to the lower nor to the upper limit. Most likely it will lie somewhere between 9 and 12 eV, provided that it is the indirect process which starts at an electron energy of 0.34 MeV. Unfortunately, both the assumption that the direct displacement process starts at 0.34 MeV and the assumption that the indirect process starts at this energy lead to reasonable values for the Sb threshold displacement energy. Therefore, we cannot decide to which of the two processes the stage-I threshold electron energy corresponds.

While there is only one stage for which the orientation dependence of the production rate indicates its relation to displaced Sb atoms, there are three stages which by the same criterion seem to involve the displacement of Ga atoms. Among these stage II is the one with by far the lowest threshold electron energy. The maximum amounts of energy which a bombarding electron can transfer to a Ga atom at the threshold electron energies of stages II–IV are 6.2, 13.7, and 14.3 eV, respectively. We interpret the lowest of these three values as the threshold displacement energy for Ga atoms and conclude that a stage-II defect is produced by the displacement of a single Ga atom.

[17] $T'_{0,\mathrm{Sb}}$ may not be equal to $T_{0,\mathrm{Sb}}$ because of the different environment of the Sb atom at the time of its displacement.

The minimum energies for creation of stage-III and stage-IV defects are about 2.2 times the minimum energy necessary to create the defects which correspond to stage II. A Ga atom having received that much energy may be capable of causing a secondary displacement. It seems plausible, therefore, that the stage-III and stage-IV defects correspond to defect configurations produced by multiple-displacement processes starting with the displacement of a Ga atom.

V. CONCLUSIONS

The results reported in the present paper lead to the following conclusions:

(1) The four major recovery stages observed in undoped GaSb after low-temperature electron irradiation correspond to the removal of defects which are produced by collisions of the bombarding electrons with regular lattice atoms.

(2) The defects corresponding to one stage are produced independently of the defects of any other stage.

(3) A stage-I defect is produced by the displacement of a single Sb atom. The threshold displacement energy for this process is larger than 7.5 eV but smaller than 13.1 eV.

(4) A stage-II defect is produced by the displacement of a single Ga atom. The Ga threshold displacement energy is about 6.2 eV.

(5) Stage-III and -IV defects are created by processes which start with the displacement of a Ga atom. Their threshold displacement energies are about 14 eV. It is conjectured that they involve secondary displacements.

ACKNOWLEDGMENTS

The author is grateful to Dr. F. H. Eisen and Dr. D. W. Keefer for a critical reading of this manuscript, and to R. A. Finch and D. D. Vawter for their technical assistance.