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#### PHYSICAL REVIEW B

#### VOLUME 5, NUMBER 8

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# Photoluminescence of Radiation Defects in Ion-Implanted 6H SiC<sup>†</sup>

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Radiation defects were introduced into 6H SiC by ion implantation and by electron bombardment. The defects produce a new low-temperature luminescence that is independent of the implanted ion, and one portion, the  $D_1$  spectrum, persists after a 1700 °C anneal. A comparison of the  $D_1$  spectrum in ion- and electron-bombarded samples shows that its intensity is strongly dependent on defect concentration, suggesting that the luminescence center is a pure-defect complex, possibly a divacancy. The  $D_1$  spectrum, which was previously observed in cubic SiC, has a strong vibronic structure with localized and resonant modes. In 6H SiC it is repeated three times, due to the three inequivalent sites in this polytype.  $D_1$  has an unusual temperature dependence, the low-temperature (1.4 °K) spectrum being extinguished as the high-temperature (77 °K) form is activated. The abrupt change of spectrum is attributed to a lattice distortion at the low temperature. The changes in luminescence on annealing are correlated with changes in electrical properties observed in ion-implanted samples by Marsh and Dunlap.

#### I. INTRODUCTION

Ion implantation in SiC results in damage to the lattice that is difficult to repair, for some defects remain even after a 1700 °C anneal. One of these persistent defects gives rise to a strong low-temperature luminescence spectrum (the  $D_1$  spectrum) that has a characteristic vibronic structure. The  $D_1$  spectrum was first studied in cubic (3*C*) SiC, and it was shown not to depend on the implanted ion.<sup>1</sup> We have now subjected 6H SiC to the same damage and annealing procedures, and we have found a very similar persistent luminescence. It can be recognized as the  $D_1$  spectrum of polytype 6H by the distinctive localized and resonant vibrational structure, but it is different from the 3Cspectrum in ways that are related to differences in crystal and band structure. For example, it is repeated three times, for there are three crystallographically inequivalent lattice sites in 6H SiC.<sup>2, 3</sup> It is also displaced to higher photon energies by about 0.6 eV, the difference between 6H and 3Cenergy gaps.<sup>4</sup>

The most striking feature of the  $D_1$  spectrum is its temperature dependence. In cubic SiC the appearance of the low- (L), medium- (M), and high-(H) temperature forms of the spectrum were tentatively attributed to Jahn-Teller distortion. The symmetry of 6H SiC is lower (space group  $P6_{3}mc$ ), and only two forms of the spectrum (called *L* and *H*) are observed. It is probable that the symmetry of the center is too low for the orbital degeneracy that leads to a true Jahn-Teller effect. <sup>5</sup> Nevertheless, the replacement of the *H* spectrum by the *L* spectrum with decreasing temperature can best be attributed to a change in lattice configuration that may be called a pseudo-Jahn-Teller distortion. <sup>5</sup> The *H* form of the spectrum was reported earlier by Makarov, <sup>6</sup> but with less resolution.

<sup>43</sup>This statement applies to the lowest-concentration

<sup>44</sup>P. P. Peressini, J. P. Harrison, and R. O. Pohl,

sample reported in Ref. 9. For the high-concentration samples dipole-dipole interactions begin to affect the

The structure of the  $D_1$  luminescence center is not known. A possible model mentioned in Ref. 1 is an impurity-vacancy pair. This choice was based on the extensive work on radiation damage and annealing in Si, where the progression from simple vacancies and interstitials to more complex defects in successive annealing stages has been well documented.<sup>7</sup> The work of Watkins and co-workers on persistent defects is especially significant.<sup>8</sup> We have found a number of sample-dependent spectra in 6H SiC that appear to be due to impurity-defect complexes. However, a comparison of ion- and electron-bombarded samples leads to the conclusion that the  $D_1$  center does not depend on the presence of an impurity. A pure-defect model

5

(e.g., a divacancy) appears to be more consistent with our experimental results. In Si the divacancy is a common and fairly stable complex.<sup>9</sup>

The electrical properties of ion-implanted 6HSiC have been studied by Marsh and Dunlap.<sup>10,11</sup> They found that the electron mobility was not restored by a 1200 °C anneal, although He back-scattering measurements indicate an almost complete recovery of the lattice structure.<sup>12</sup> They also had difficulty in making a p-type layer by ion implantation, as if a compensating defect were present. We have observed luminescence changes following annealing that can be tentatively correlated with the electrical changes. Various impurity-defect spectra disappear in the annealing range 1100-1500 °C in which Marsh and Dunlap found an increasing mobility in n-type implants. The  $D_1$ spectrum persists, however, and one can speculate that  $D_1$  is the compensating center that impedes p-type implants.

The *L* and *H* forms of the  $D_1$  spectrum at 1.4 and 77 °K are shown in Sec. III. All *L* and *H* nophonon lines are shown together at 22 °K in Sec. IV, and the activation energies and pre-exponential factors are tabulated. The  $D_1$  centers in 6*H* and 3*C* polytypes are compared in Sec. V, and the effects of crystal symmetry are considered. Information on impurity-defect complexes obtained from electron-bombarded samples is given in Sec. VI, and the annealing changes in luminescence and electrical properties are discussed in Sec. VII.

### **II. EXPERIMENTAL PROCEDURES**

Samples used for the spectra shown here were bombarded with He ions  $(5 \times 10^{14} \text{ cm}^{-2} \text{ at } 150 \text{ keV})$ , using an ion accelerator equipped with beam scanner and magnetic separator. The unchanneled penetration of the He ions was about 0.65  $\mu$ m.<sup>13</sup> For exciting the luminescence, the penetration of the Hg light was kept to about the same distance by using filters to exclude wavelengths greater than 3100 Å. This reduced the background luminescence from the unbombarded crystal.

He ions were used because of their relatively deep penetration, and because we wished to implant a neutral ion that would not itself contribute to the luminescence spectrum. That the He ions play no part in the luminescence was confirmed by the observation of the same spectrum in samples bombarded with Ne, Ar, Ag, and I ions. The spectrum was also observed in samples bombarded with 1-MeV electrons  $(6 \times 10^{17} \text{ cm}^{-2})$  in a Van de Graaff accelerator.

Most of the spectra were recorded on Kodak 103F plates, from which densitometer traces were made. For accurate intensity ratios we used a different system that employed a photomultiplier detector.

All annealing at T > 800 °C was done in an Ar

atmosphere. The annealing time was 15 min for T < 1600 °C, but was reduced to 2 min for  $T \ge 1600$  °C to avoid extensive thermal etching of the samples.<sup>14</sup>

#### **III. EXPERIMENTAL RESULTS**

Figure 1 shows a portion of the 1.4 °K luminescence spectrum of a 6H sample bombarded with He ions  $(5 \times 10^{14} \text{ cm}^{-2})$  and then annealed at 1300 °C. This is the same treatment given to the cubic SiC in Ref. 1.  $L_1$ ,  $L_2$ , and  $L_3$  are the three low-temperature no-phonon lines corresponding to the three inequivalent sites of 6H SiC. Most of the vibronic components in Fig. 1 belong to  $L_1$ , and the zero of the upper-energy scale begins at  $L_1$  to facilitate the reading of phonon energies. The vibronic peak at the gap in the scale indicates that a localized mode of  $L_1$  falls between the bands of acoustic and optic lattice phonons. Phonon structure due to  $L_2$  or  $L_3$ is marked 2 or 3. Some weak but sharp extraneous lines have been suppressed in Fig. 1. They were found to anneal out at about 1500 °C.

Beyond the one-phonon cutoff of  $L_1$ , in a portion of the spectrum not shown, there is a great deal more structure in which many of the  $L_2$  and  $L_3$  vibronic components are resolved. They are less distinct than the  $L_1$  components because they are weaker and because they overlap each other and the two-phonon structure of  $L_1$ . However, all the well-resolved peaks indicate that the three centers have essentially the same vibronic structure.

Figure 2 shows the spectrum of the same crystal



FIG. 1. Portion of the  $D_1$  luminescence spectrum of an ion-bombarded 6H sample at 1.4 °K, after a 1300 °C anneal. The lines  $L_1$ ,  $L_2$ , and  $L_3$  are the three low-temperature no-phonon lines. Most of the vibronic bands in this portion belong to the  $L_1$  part of the spectrum. The gap in the phonon-energy scale is the gap between acoustic and optic lattice modes.



FIG. 2. Same spectral region and sample as in Fig. 1, but measured at 77 °K. The lines  $H_1$ ,  $H_2$ , and  $H_3$  are the three high-temperature no-phonon lines. The vibronic structure is not as well resolved at this temperature.

at 77 °K.  $H_1$ ,  $H_2$ , and  $H_3$  are the three high-temperature no-phonon lines that replace the *L* lines. Very faint *L* lines can still be detected on the photographic plate but do not show on the densitometer trace. With respect to the *L* lines, the  $H_1$ ,  $H_2$ , and  $H_3$  lines are displaced to higher energies by 11.3, 10.7, and 22.1 meV, respectively.

As in Fig. 1, the spectrum is cut off at the end of the  $H_1$  one-phonon vibronic components. In this portion no  $H_2$  or  $H_3$  phonon structure is clearly resolved, although  $H_2$  and  $H_3$  contribute significantly to the region marked LO. Beyond the Fig. 2 cutoff, where  $H_2$  and  $H_3$  components are resolved, they duplicate the  $H_1$  components, which, in turn, duplicate with somewhat less resolution the vibronic components of the three L spectra. Thus, the same vibronic spectrum is repeated six times altogether. We shall use  $D_1$  as a comprehensive name for the luminescence center or for the complete set of spectra regardless of temperature or polytype. For example, we say that L and H are the low- and high-temperature forms of the  $D_1$  spectrum in 6HSiC.

The six most prominent phonon contributions to these vibronic spectra are shown in Table I. Three of the phonons can be identified with known lattice phonons of 6H SiC, <sup>2, 15</sup> which are, in fact, nearly identical in all polytypes, as shown by Raman scattering. <sup>16</sup> The other three phonons can be identified as damage-induced localized and resonant phonons of the  $D_1$  center also observed in ion- and electron-bombarded cubic SiC.<sup>1</sup> The mode falling in the gap between acoustic and optic branches is localized. The in-band modes are resonant, <sup>17</sup> and in Ref. 1 it was shown that they are strongly displaced from maxima in the phonon density of states.

The identification of  $D_1$  as the same defect in 6Hand 3C polytypes was suggested by the unusual temperature dependence of its spectrum and by its resistance to annealing, but is more firmly based on the characteristic localized and resonant modes. There could not be similar vibronic structure in the two polytypes unless there were also similar phonon densities of states. Thus, the vibronic spectra are like the Raman scattering data<sup>16</sup> in emphasizing that polytype differences in phonon structure are often negligible. We ignore the fact that 6H SiC, with 12 atoms per unit cell, has thirtysix phonon branches, and we use the standard large zone to reduce the number to six branches with the approximate but familiar notation (TA, LA, TO, and LO).  $^{15}$ 

If we consider the luminescence to be due to exciton recombination at the  $D_1$  center, we find that the energies binding excitons to the centers are comparable in 3C and 6H polytypes. We take the difference between the photon energy of the no-phonon line and the energy gap, 3.023 eV in 6H, or 2.390 eV in 3C, <sup>4</sup> to obtain binding energies of 0.398, 0.434, and 0.453 eV for the three 6H centers, and 0.416 eV for the 3C center. These values are for the low-temperature forms of the  $D_1$  complexes. The binding energies of excitons to the high-temperature forms are reduced by the L to H activation energies that are discussed in Sec. IV.

# IV. RESULTS AT INTERMEDIATE TEMPERATURES

In Fig. 3 we show the no-phonon region of the  $D_1$  spectrum at 22 °K. At this temperature the three L and the three H no-phonon lines appear together, although  $H_3$  is weak because of its high activation energy. The activation energies  $E_i$  (i = 1, 2, 3) are taken from the  $H_i$ - $L_i$  line separations on the assumption that the luminescence transitions are to the same final state for both forms of the center. The identification of H and L pairs is fairly clear from Fig. 3, but it was confirmed by taking the

TABLE I. Six prominent vibrational modes observed in the  $D_1$  spectrum of 6H SiC, and their identification as lattice modes or as localized and resonant modes.

Mode	Energy (meV)	Lattice (6 <i>H</i> )	D <sub>1</sub> spectrum in 3C
Resonant LA	68		67
Zone edge LA	76	76	
Localized (Gap)	83		83
Zone center TO	98	98	
Resonant LO	110 - 112		110 - 113
Zone center LO	120	120	



FIG. 3. No-phonon lines of the  $D_1$  spectrum at an intermediate temperature (22 °K), showing both L and H, the low- and high-temperature forms. The energy and wavelength are shown for each of the six lines. A photomultiplier was used as detector to obtain accurate intensity ratios of the pairs,  $H_t/L_t$ .

intensity ratios  $H_i/L_i$  at several temperatures, <sup>18</sup> and fitting these ratios to the three expressions

$$H_i/L_i = c_i e^{-E_i/kT}, \quad i = 1, 2, 3.$$
 (1)

The pre-exponential factors  $c_i$  and activation energies  $E_i$  are shown in Table II, where the comparable data for 3*C* SiC are also given.

The transition from L to H in 3C SiC goes through the intermediate M spectrum (M for medium-temperature range). Nevertheless, the 3C pre-exponential factor for the H/L intensity ratio is comparable with one of the 6H factors ( $c_3$ ). The reason for the differences in the three values of  $c_i$  in 6H SiC is not understood, for our knowledge of the center and its distortion is very limited.

The total activation energy for the two 3C transitions is only 3.9 meV (0.7 meV for L to M and 3.2 meV for M to H). The much larger activation energies for the 6H transitions suggest that the lower 6H symmetry may permit a larger low-temperature distortion of the  $D_1$  center. These large values of  $E_i$  also give us another strong argument against a j-j coupling model, in addition to that

TABLE II. Activation energies  $E_i$  (i = 1, 2, 3) and pre-exponential factors  $c_i$  used in fitting Eq. (1) for 6*H* and 3*C* intensity ratios  $H_i/L_i$ .

	c ;	E <sub>i</sub> (meV)
$H_1/L_1$	270	11.3
$H_2/L_2$	350	10.7
$H_3/L_3$	1400	22.1
H/L (3C)	1500	3.9

already given in Ref. 1. The differences between 3C and 6H centers are thought to be closely related to the question of symmetry and will be considered next.

## V. SYMMETRY OF 6H AND 3C CENTERS

Subjecting 6H and 3C SiC to the same damage and annealing procedures produces the same persistent luminescent center, as judged by the localized and resonant vibrational modes. To account for the 6Hand 3C differences we need a model of the center, and we reduce the many possibilities to those we consider most likely, namely, a vacancy-impurity pair, or a divacancy on adjacent lattice sites.<sup>19</sup> We use the vacancy-impurity pair to show the symmetry of the center in Fig. 4, but the substitution of a second vacancy for the impurity serves to illustrate the divacancy, and leaves the model symmetry unchanged. Either center has  $C_{3\nu}$  symmetry in cubic SiC, but in 6*H* the symmetry is  $C_{3\nu}$  only if the vacancy-impurity or vacancy-vacancy pair is lined up along the c axis. The nonaxial center, shown in Fig. 4, has only  $C_{1h}$  symmetry.

The 6H axial center should be very similar to that of 3C SiC. A threefold-degenerate distortion would be expected on the Watkins-Corbett model, <sup>7-9</sup> and this would apparently lead to three forms of the center with increasing temperature, namely, the



FIG. 4. Model of a nonaxial vacancy-impurity complex. The vacancy is at position 2 and the impurity (I) at position 3. The neighbors of the vacancy, 5 and 6, are shown forming a bond, leaving the c directed and therefore inequivalent atom 1 to form the luminescence center with the impurity. Substitution of another vacancy for the impurity provides a model of the nonaxial divacancy.

static, dynamic, and undistorted forms, as in 3C.<sup>1</sup> However, the nonaxial center permits a unique distortion which necessarily goes from the static form directly into the undistorted form at a higher temperature, bypassing the dynamic range to which we attribute the *M* spectrum in 3C. Figure 4 illustrates how this distortion may occur by the formation of a bond joining atoms 5 and 6, leaving the orbital electrons of the center primarily on atoms 1 and 3. Atom 1 is slightly different from atoms 5 and 6 because its direction from the vacancy is along the *c* axis.

If the model is appropriate, it seems likely that only the nonaxial centers contribute to the 6H luminescence. Figure 4 shows that a nonaxial center is converted into an axial center when atom 4 jumps into the vacancy 2. Presumably the nonaxial center is one of lower energy and therefore is the only active center throughout the temperature range of our measurements.

As we remarked, it may be the lower  $C_{1h}$  symmetry of the nonaxial center that accounts for the large L-to-H activation energies, > 10 meV, compared with the 3.9 meV observed in 3C. However, only nondegenerate electronic states are possible for a  $C_{1h}$  center (excluding spin degeneracy), so that a Jahn-Teller effect cannot occur.<sup>5</sup> Probably the presence of a vacancy makes the low-temperature distortion energetically favorable, resulting in what is sometimes called a pseudo-Jahn-Teller effect. This distortion occurs in the excited state of the center, before photon emission.

#### VI. ELECTRON-BOMBARDED SAMPLES

The  $D_1$  spectrum is also observed in samples that have been bombarded by electrons  $(6 \times 10^{17} \text{ cm}^{-2}$ at 1 MeV). The no-phonon lines can be detected after a 700 °C anneal, but are weaker than competing spectral lines.  $D_1$  does not become dominant until the annealing temperature reaches about 1400 °C, with some sample dependence. This is in contrast with ion-bombarded samples, in which  $D_1$  is dominant at 1000 °C, the lowest annealing temperature used. The electron-bombarded samples serve two purposes, for (i) they permit us to study the low-temperature annealing stages and (ii) they show the effects of a comparatively low defect concentration.

A 1-MeV electron may travel 1 mm in SiC and produce one vacancy-interstitial pair, whereas a 150-keV He ion travels less than 1  $\mu$ m and produces several hundred defect pairs. With the fluences used we produced approximately the same number of vacancy-interstitial pairs in ion- and electron-bombarded specimens, but the defects in the ion-bombarded samples are confined to the 1- $\mu$ m surface layer, hence are concentrated by a factor 10<sup>3</sup>. Consequently, the heavily damaged layer of an ion-bombarded sample must be annealed to about 1000 °C to produce a luminescence spectrum with good structure, at which temperature the  $D_1$  spectrum is already dominant. In contrast, the electron-bombarded samples require no anneal.

It is instructive to study the larger range of annealing temperatures in the electron-bombarded samples. We observe numerous annealing steps, as demonstrated by the changing luminescence spectra. In the annealing range from room temperature to 1100 °C we see considerable sampleto-sample differences in the spectra, indicating the involvement of impurities in many of the luminescence centers. The differences are very obvious if we compare samples from *different* crystalgrowth furnace runs. Often the spectra of such samples have many lines in common, but the relative intensities of the lines are quite different. It seems clear that such sample dependence should be attributed to differences in the various impurity densities. Thus, many of the luminescence centers require both crystal defects and impurities for their existence, and may be called impurity-defect complexes. A probable center, in view of the Si work.<sup>7-9</sup> is an impurity-vacancy pair. The different impurities provide vacancy traps of different depths, thus giving rise to the numerous annealing stages. However, the  $D_1$  spectrum appears in *all* our samples, and also in the samples used by Makarov, <sup>6</sup> so we cannot say that an impurity is a necessary part of this center.

The most striking difference between electronand ion-bombarded samples is the much greater strength of the  $D_1$  spectrum in the latter. In order to compare samples from the *same* crystal-growth furnace run, we show in Fig. 5 the many spectral lines in a portion of the 1.4 °K luminescence of an electron-bombarded sample from the same run as the ion-bombarded sample of Figs. 1-3. The annealing temperature was also the same, 1300 °C.<sup>20</sup> The  $D_1$  spectrum is present in Fig. 5, as indicated by the lines  $L_1$ ,  $L_2$ , and  $L_3$ , but  $D_1$  does not stand out with respect to the many impurity-defect lines. In contrast, impurity-defect lines were only faintly visible in Fig. 1. The very different relative strength of the  $D_1$  spectrum indicates the strong dependence on defect concentration that one might expect for a pure-defect complex.

In Si the divacancy is a common and relatively stable pure-defect complex, <sup>21</sup> although its annealing temperature of about 250 °C is below that of some impurity-defect complexes in Si, and *far* below that of the  $D_1$  center in SiC. We may consider the C-Si nearest-neighbor divacancy as a possible model for the  $D_1$  center. Its formation probability would be proportional to the square of the defect concentration; hence its prominence in the ion-bombarded





FIG. 5. Some of the luminescence lines in an *electron*bombarded 6H SiC sample at 1.4 °K, after a 1300 °C anneal. Comparison with Fig. 1 shows the great difference between ion- and electron-bombarded specimens. The pure-defect lines  $L_1$ ,  $L_2$ , and  $L_3$  are no more prominent here than the numerous impurity-defect lines.

samples would be understandable, and its appearance in *all* samples a natural consequence.

## VII. ANNEALING OF LUMINESCENCE CENTERS

Examination of the luminescence spectra, after successive annealing stages, permits us to correlate changes in luminescence with changes in electrical properties observed in the annealing experiments of Marsh and Dunlap.<sup>10</sup> The presence of sharp lines in the luminescence makes it possible to see which defects anneal out at each stage. However, the luminescence observations are not very quantitative, for absolute intensity measurements are difficult and most of the excitation energy goes into competing processes, both radiative and nonradiative, that also change with annealing. We may conveniently divide the annealing range into two parts, 1100–1500 and 1500–1700 °C.

In the range 1100–1500 °C we note the disappearance of the impurity-defect complexes. The puredefect  $D_1$  remains, but the lines become narrower, perhaps because of the annealing of the other defects. In this range Marsh and Dunlap found a steady increase of the electron mobility in *n*-type implants, although little change was evident in He back-scattering measurements.<sup>12</sup> Thus, the luminescence measurements suggest that impuritydefect complexes are also effective scattering centers for electrons.

In the range 1500–1700 °C,  $D_1$  persists, with the intensity decreasing by about a factor of 3. However, thermal etch pits were very evident on the thin implanted layer after the 1700 °C anneal. Since we cannot tell whether the loss of  $D_1$  intensity is due to the partial annealing of  $D_1$  or to the etching of the surface, no further annealing was undertaken. However, we can state that  $D_1$  remains until 1700 °C at least. If  $D_1$  is a divacancy it may have several charge states like the Si divacancy.<sup>9</sup> The latter is positively charged when the Fermi level approaches the valence-band edge, and if the same is true of the SiC divacancy, it would act as a compensating center in any attempt to implant acceptors. Thus, the persistence of  $D_1$  is a possible explanation of the failure of Marsh and Dunlap to obtain good ptype implants.

#### VIII. SUMMARY

New luminescence centers are created in 6H SiC by ion- or electron-bombardment. Many of these are sample dependent, and can be attributed to impurity-defect complexes, possibly impurity-vacancy pairs. However, the  $D_1$  center, which resists annealing at 1700 °C, appears to be a pure-defect complex, possibly a divacancy. The  $D_1$  spectrum is similar to one reported for cubic SiC with respect to its formation, resistance to annealing, and vibronic sidebands. However, the 6H crystal structure leads to a threefold multiplicity and to the omission of one of the temperature-dependent stages. The latter is attributed to the lower symmetry of the 6H polytype, for the  $D_1$  center in 6Hprobably does not have the orbital degeneracy necessary for a true Jahn-Teller effect. A low-temperature distortion may occur nevertheless, leading to what is sometimes called a pseudo-Jahn-Teller effect.

The impurity-defect complexes anneal out in the range 1100-1500 °C in which electrical measurements on *n*-type implants show an increase in mobility. The  $D_1$  center resists annealing even at 1700 °C and may be the center that compensates acceptors in *p*-type implants.

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## PHYSICAL REVIEW B

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VOLUME 5, NUMBER 8

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# Radiation Annihilation of *F*-Aggregate Centers in KCl<sup>†</sup>

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The radiation annihilation of F-aggregate centers has been measured in KCl as a function of trace impurity and of temperature. The data have been found to support the idea that Faggregate-center destruction occurs when a radiation-produced interstitial is trapped by an aggregate center. Comparison of the results with the F-center-production behavior of the same material has permitted the following conclusions about *F*-center production to be drawn: (i) The impurity dependence of F-center production at 80  $^{\circ}$ K is due to a secondary process involving motion of interstitials. (ii) The large rise with temperature of the F-center-production efficiency between 120 and 240  $^{\circ}$ K is a reflection of the temperature dependence of the primary defect-production mechanism and may be related to a change of the relative probabilities of Frenkel-pair separation and recombination during the production process. Analysis of the experimental results has also yielded values for the energy necessary to produce a Frenkel pair at various temperatures. Above 200 °K this energy is so low [less than 100 eV/(defect pair)] that it is possible to rule out defect-production mechanisms that require multiple ionization.

#### I. INTRODUCTION

In alkali halides many of the more simple lattice defects produced by ionizing radiation, particularly anion vacancies and vacancy aggregates, have been identified.<sup>1,2</sup> Our research interests are now centered on the processes involving these defects, such as production, aggregation, and annihilation.<sup>3</sup> In this paper, measurements of F-aggregatecenter decay during electron irradiation at temperatures between 80 and 250 °K are described. The results are analyzed to yield detailed information about the processes of  $F_2$ -center<sup>4</sup> annihilation and F-center production. The  $F_2$  center, consisting of two adjacent-anion vacancies containing two electrons, gives rise to a well-known

and well-resolved optical-absorption band which in KCl peaks at 800 nm. The height of this band is proportional to the concentration and therefore lends itself to determining  $F_2$ -center-concentration changes.

The fact that  $F_2$  centers and F centers are in equilibrium during irradiation at room temperature was recognized a number of years ago.<sup>5</sup> It was discovered then that this equilibrium was influenced by dose rate and temperature. Clearly, there must be aggregation of F centers and a reaction to either break up or destroy the aggregates. From a number of studies<sup>6-9</sup> the conclusion has been reached that the  $F_2$  centers form when F centers are ionized and the resulting  $F^+$  centers move. Little work has been done to identify the reaction