Damage production and annealing in $^{28}$Si-implanted CoSi$_2$/Si(111) heterostructures

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The damage in epitaxial CoSi$_2$ films 500 nm thick grown on Si(111) produced by room-temperature implantation of 150 keV $^{28}$Si were investigated by 2-MeV $^4$He channeling spectrometry, double-crystal x-ray diffractometry, and electrical resistivity measurements. The damage in the films can be categorized into two types. In lightly (heavily) damaged CoSi$_2$ the damage is in the form of point-like (extended) defects. The resistivity of lightly damaged CoSi$_2$ films rises with the dose of implantation. Electrical defects correlate well with structural ones in lightly damaged films. The resistivity of heavily damaged films flattens off while the structural defects continue to rise with the dose, so that resistivity no longer correlates with structural defects. Upon thermal annealing, lightly damaged films can fully recover structurally and electrically, whereas heavily damaged films do so only electrically. A residual structural damage remains even after annealing at 800 °C for 60 min.

I. INTRODUCTION

Transition-metal silicides are widely used as contacting layers in the metal contacts of Si-based electronic devices. The damage produced by the implantation of dopants in a silicide layer and its recovery thus have technological importance. Compared to Si, the metallic silicides are highly resistant to radiation damage. Most previous studies have focused on the amorphization of the silicides by implantation and the recrystallization by subsequent thermal annealing. An amorphized CoSi$_2$ layer on a crystalline seed recrystallizes in a layer-by-layer manner by solid phase epitaxy. Hensel et al. studied the effect of ion implantation on carrier transport of CoSi$_2$ films and found that the resistivity increases with dose. We report here some studies on defects production and their annealing in room-temperature $^{28}$Si implanted CoSi$_2$ films. The emphasis is on the strain, the damage, and their relationship with the defects in the implanted films. The correlation between structural and electrical defects is also investigated.

II. EXPERIMENTAL PROCEDURES

An epitaxial CoSi$_2$ film 50 nm thick was grown on Si(111) at ~600 °C by molecular beam epitaxy. The samples were implanted at room temperature in vacuum ($\sim 10^{-7}$ Torr) by 150-keV $^{28}$Si ions to doses from $5 \times 10^{13}$/cm$^2$ to $3 \times 10^{15}$/cm$^2$. The beam flux was kept at ~0.2 µA/cm$^2$ to minimize the sample heating. The direction of the incident beam was ~7° away from the [111] axis to avoid channeling. The maximum damage is located in the Si substrate at a depth of ~150 nm beneath the interface, according to a TRIM88 simulation. The damage in the implanted layers was characterized by 2-MeV $^4$He ion channeling, Fe Kα$_1$ x-ray (wavelength $\lambda = 0.1936$ nm) double-crystal diffraction, and electrical resistivity measurements.

The implanted samples were annealed afterwards in vacuum ($\sim 5 \times 10^{-7}$ Torr) at 250–800 °C for 60 min and the damage recovery was monitored.

III. RESULTS

A. MeV $^4$He channeling spectrometry

2-MeV $^4$He[111] axial channeling with a glancing exit angle (82°) was used to measure the damage in the implanted layers. The spectra of the implanted samples show that the channeling yields of both the film and the substrate rise with the dose (Fig. 1). For a given dose, the damage level in the CoSi$_2$ film is much smaller than that in the Si substrate, meaning that CoSi$_2$ is more radiation resistant than Si. The high radiation resistance of CoSi$_2$ agrees with the results of room-temperature implanted Pd$_2$Si and NiSi$_2$, and is consistent with the metallic nature of these silicides.

The damage build up in the Si substrate as a function of the dose is very similar to that observed in self-implanted Si samples. There exist three distinct damage regimes (see Fig. 1). The damage rises slowly as the dose increases from $5 \times 10^{13}$/cm$^2$ to $2 \times 10^{14}$/cm$^2$ (regime I), and then rapidly rises to the level of an amorphous Si at $5 \times 10^{14}$/cm$^2$ (regime II). As the dose increases further, the maximum damage level saturates, and the damage region widens (regime III).

The damage in the CoSi$_2$ film builds up very differently with dose. Below $2 \times 10^{14}$/cm$^2$ the channeling yield of the film is the same as that of the as-grown one (~5%). The channeling yield increases to ~0.2% at ~0.1% at $5 \times 10^{14}$/cm$^2$, and rises to ~56% at $5 \times 10^{14}$/cm$^2$ (Fig. 1). Maex et al. found that the channeling yield of CoSi$_2$ films implanted at room temperature by 200-keV $^{28}$Si to $2 \times 10^{15}$/cm$^2$ is ~50%, in good agreement with our results. They also observed that...
The damage in CoSi$_2$ is sensitive to the implantation temperature, as it is in Si.\cite{13} Hewett \textit{et al.}\cite{5} demonstrated that CoSi$_2$ can be amorphized by 40–200-keV $^{28}$Si implantation to $\sim 2 \times 10^{15}$/cm$^2$ at liquid-nitrogen temperature.\cite{5-8}

We apply the procedure outlined in Ref. 12 to estimate the defect concentration at a depth $x$, $c_D(x)$ ($0 < c_D < 1$), in the implanted CoSi$_2$ film from the channeling measurements. To simplify the analysis, we assume that the fraction of displaced Si atoms in the film is the same as that of displaced Co atoms. The dechanneling factor, $\gamma_D(x)$, is obtained directly from the channeling spectra,\cite{12}

$$\gamma_D(x) = \frac{\chi_D(x) - \chi_P(x)}{1 - \chi_P(x)} \quad (1)$$

where $\chi_D$ and $\chi_P$ are the channeling yield of the damaged (virgin) sample normalized to the random yield. The defect concentration profile $c_D(x)$ can be extracted from the equation,\cite{12}

$$c_D(x) = P_D(x) + [1 - P_D(x)]c_P(x) \quad (2)$$

$P_D(x)$ is the dechanneling probability at a depth $x$.\cite{12}

$$P_D(x) = \int_0^x n\sigma_D c_D(x')dx' \quad (3)$$

where $n$ is the atomic density of CoSi$_2$, and $\sigma_D$ is the average dechanneling cross section of Si and Co atoms. TRIM88 simulation shows that the damage is roughly a constant through the entire film. The channeling spectra of Fig. 1 suggest the same conclusion. We therefore assume that $c_D$ is depth-independent in the film, and Eq. (3) becomes

$$P_D(x) = (n\sigma_D c_D(x) \quad (4)$$

We compute $\gamma_D$ from the minimum channeling yield of the Co signal at the energy immediately beneath the surface peak ($\sim 1.65$ MeV in Fig. 1). The corresponding depth is $x \approx 10$ nm. Substituting the appropriate numbers into Eq. (4), and assuming $\sigma_D < 10^{-18}$/cm$^2$ (see Ref. 12), one obtains

$$P_D < 0.05 c_D$$

$P_D$ is small compared to $c_D$, and we therefore neglect the second term in Eq. (2) and have

$$c_D \approx \gamma_D$$

(5)

The defect concentration $c_D$ in the CoSi$_2$ films, extracted from the channeling data according to Eq. (5), is plotted as a function of the implantation dose in Fig. 2.\cite{12} $c_D$ is undetectable ($<1$%) after the implantation to $2 \times 10^{14}$/cm$^2$, and then increases monotonically with the dose.

For comparison, we also computed by a TRIM88 simulation\cite{11} the concentration of Frenkel pairs directly produced by the implantation of 150-keV $^{28}$Si in an amorphous matrix composed of a 50-nm-thick CoSi$_2$ on Si (dashed line in Fig. 1). A displacement threshold energy of 15 eV and a binding energy of 1 eV were assumed. The measured damage is only a small fraction of that initially produced, meaning that the majority of the Frenkel pairs anneal out at room temperature during and after implantation. This observation leads us to attribute the high radiation resistance of CoSi$_2$ to its high defect recombination rate, probably resulting from the higher mobility of point defects in CoSi$_2$ than in Si.

Upon post-thermal annealing, the damage in the Si substrates and CoSi$_2$ films decreases gradually. The annealing characteristics of the damaged Si substrate is similar to that of implanted bulk Si.\cite{12} Different damage regimes have different recovery stages. The higher the damage, the higher the annealing temperature becomes. At low damage levels ($\phi < 2 \times 10^{14}$/cm$^2$, regime I), the channeling yield decreases as the annealing temperature rises. The dominant process is probably the recombination of point-like defects. At high damage levels near the amorphization threshold ($\phi = 5 \times 10^{14}$/cm$^2$, regime II), the channeling spectrum does not change after 60-min isochronal annealing at 250°C. The channeling yield decreases after 400°C annealing, and becomes the same as that of the as-grown sample after 600°C annealing. The channeling spectrum of the amorphized samples ($\phi > 10^{15}$/cm$^2$, regime III) does not change after 400°C annealing. In that case, appreciable solid phase epitaxial growth occurs after annealing at 600°C. However, the channeling yield is very high (50%) and remains that high after 800°C annealing. A high density of extended defects (e.g., dislocation loops, microtwins) is likely present.

Qualitatively, the annealing features of the CoSi$_2$ films are simpler than those of the Si substrate, since the film is not amorphized after implantation to the highest dose of $3 \times 10^{14}$/cm$^2$ (see Fig. 1). The channeling yields of the implanted films decrease with increasing annealing temperature. The defect concentration in the film after annealing at various temperatures, extracted from the channeling yield according to Eq. (5), is shown in Fig. 2. The films implanted to a dose of $< 5 \times 10^{14}$/cm$^2$ completely recover.
while those implanted to doses \( \geq 10^{15}/\text{cm}^2 \) do not. We believe that the nonzero defect concentration (\(-10\%\), Fig. 2) after annealing is indicative of the presence of residual defects that are extended, rather than a measure of the concentration of point defects. These observations suggest that there exist two distinct damage structures in ion implanted CoSi, The defects in the lightly damaged films (\( \phi < 5 \times 10^{14}/\text{cm}^2 \)) are simple and anneal out completely upon thermal processing. The defects in the heavily damaged films (\( \phi \geq 10^{15}/\text{cm}^2 \)) are complex and the recovery is incomplete. We note that the incomplete recovery of the film coincides with that of the Si substrate.

B. X-ray double-crystal diffractometry

A heterostructure differs from a bulk crystal in that there exists intrinsic strain in an as-grown heterostructure. Implantation produces defects, which induce additional strain in the heteroepitaxial film. We use x-ray double-crystal diffraction to monitor the strain change in the CoSi, film as a result of the implantation. Figure 3 shows a set of x-ray rocking curves diffracted from the (111) symmetrical plane of the CoSi,/(Si(111)) samples implanted to a dose of (a) 0X, (b) 0.5X, (c) IX, (d) 2X, (e) \( 5 \times 10^{14}/\text{cm}^2 \).

Based on previous studies on implantation in Si,(100),\(^{12}\) we expect that the damage in the Si substrate induces positive strain, and that an additional diffraction peak will appear on the low-angle side of the bulk Si peak at \( \theta_B = 18^\circ \) in Fig. 3. Implantation to \( \phi < 2 \times 10^{14}/\text{cm}^2 \) produces low damage (regime I) in the Si substrate, and induces a small perpendicular strain of \(<0.1\%\) there (see Ref. 12). The shift in the angular position for the (111) diffraction peak of the implanted layer is correspondingly small (\(<0.02^\circ\)) and undetectable because it lies within the main peak (see Fig. 3). Implantation to \( \phi > 2 \times 10^{14}/\text{cm}^2 \) produces high damage (regime II) or a continuous amorphous layer (regime III)\(^{12}\) the corresponding diffraction peak then becomes weak and is buried in the background intensity (\(-0.003\%), see Fig. 3).

The peak intensity of x-rays diffracted from the CoSi, (111) planes decreases with increasing dose (Fig. 3). After implantation to a dose of \( 2 \times 10^{14}/\text{cm}^2 \), the intensity drops to \(~1\%) of that of the as-grown film, unlike the channeling yield which is the same as that of the as-grown sample (Figs. 1 and 2). This fact demonstrates that x-ray diffraction is much more sensitive to low defect concentrations than channeling. The x-ray intensity from the CoSi, layer of the sample implanted to a dose of \( >10^{15}/\text{cm}^2 \) drops below the background (\(-0.003\%\)) and hence becomes unmeasurable. These two techniques are therefore complementary and together provide a good picture of damage production by ion implantation in CoSi,.

In the following, we will analyze quantitatively the x-ray rocking curves diffracted from the lightly damaged CoSi, films (\( \phi < 5 \times 10^{14}/\text{cm}^2 \)). The angular position of the (111) diffraction peak does not change with the dose, which means that the perpendicular strain in the films is a constant, \( \sim -1.76\% \), within the experimental sensitivity (\( \sim 0.08\% \)). In other words, the upper limit in the difference of the magnitude of the strain between the implanted and as-grown films, \( \Delta C \), is \( \sim 0.08\% \). The defect concentration \( C_D \) in the corresponding films ranges from \( \sim 1\% \) to \( \sim 4\% \) (\(\bullet\) in Fig. 2). These results imply that in the lightly damaged CoSi, films, one has

\[ |\Delta C| < 0.02C_D.\]

We were unable to measure the parallel strain in our CoSi, films because they are B type.\(^{14}\) We know, however, that the interfacial misfit dislocations do not shear at room
temperature. It is therefore reasonable to assume that the parallel strain of the film remains unchanged upon the implantation, i.e., $\Delta \varepsilon^p = 0$. It then follows (see Ref. 12) that the lattice dilatation, $\Delta a/a$, of the CoSi$_2$ induced by the defects is

$$
\frac{\Delta a}{a} = \frac{1 - \nu}{1 + \nu} |\Delta \varepsilon| < 0.01 c_{pd},
$$

where $a$ is the lattice constant of the as-grown CoSi$_2$ film, and $\nu$ is Poisson’s ratio of CoSi$_2$ ($=\frac{1}{3}$ see Ref. 15). The coefficient relating $|\Delta a/a|$ to $c_{pd}$ for ion implanted CoSi$_2$ is $<0.01$. This value is about two orders of magnitude smaller than that for vacancy-interstitial pairs in fcc crystals ($\sim 1$). but is about the same as that for ion implanted Si ($\sim 0.007$, see Ref. 12). This fact suggests that the defects in lightly damaged CoSi$_2$ films consist of defect clusters, not of single vacancies and interstitials. This result is consistent with our previous assertion that point defects in CoSi$_2$ are mobile at room temperature.

The full width at half maximum of all diffraction peaks that are strong enough to be detected (films implanted to $< 5 \times 10^{14}$/cm$^2$) is an invariant (see Fig. 3), independent of the damage induced by ion implantation. The peak broadening in the present case is due to the finite film thickness. This result implies that the x-ray diffraction from the lightly damaged film is highly coherent, and suggests that the damage in the implanted film consists mainly of randomly distributed defect clusters. The x-ray diffraction from a lightly damaged CoSi$_2$ film can therefore be modeled by a static Debye-Waller factor, which does not change the peak width but only decreases the intensity.

The ratio of the peak intensity from the implanted film, $I_D$, to that from the as-grown one, $I_V$, equals,

$$
\frac{I_D}{I_V} = \exp \left( -\frac{16\pi^2 \sin^2 \theta_B}{3 \lambda^2} \right),
$$

where $u_B$ is the mean square root of the atomic displacement caused by the point-like defects in the implanted film. For Fe $K_{al}$ x rays diffracted from CoSi(111) planes, Eq. (8) results in

$$
u = 0.085 \sqrt{\ln \frac{I_V}{I_D}} \text{ (nm)}.\tag{9}
$$

The displacement $u_B$ in the lightly damaged CoSi$_2$ film, extracted from the measured intensities in Fig. 3 according to Eq. (9), rapidly rises to $\sim 0.06$ nm after implantation to $5 \times 10^{13}$/cm$^2$ and then increases linearly with dose (● in Fig. 4).

Upon 30-min isochronal annealing at 250, 400, 600, and 800 °C in high vacuum, both the angular position and the width of the x-ray diffraction peak from the lightly damaged CoSi$_2$ films ($c_{pd} < 4\%$) remain unchanged. This means that within the experimental sensitivity, the strain of the CoSi$_2$ films does not change, $\varepsilon \sim 1.76\%$ at room temperature. The presence of a small defect concentration in the metastable CoSi$_2$ film (a 50-nm-thick CoSi$_2$ film on Si(111) at an equilibrium state has $\varepsilon \sim 1.23\%$ at room temperature) during thermal annealing evidently does not enhance the strain relaxation. The x-ray peak intensity rises with increasing temperature. For the films implanted up to $2 \times 10^{14}$/cm$^2$, the intensity after $250\, ^\circ\text{C}$ is the same as that of the as-grown film within the experimental uncertainty ($\sim 20\%$). The uncertainty in the intensity measurements causes a corresponding uncertainty in the estimation of the atomic displacement extracted according to Eq. (9),

$$
\frac{\delta u_D/u_D}{\sqrt{(\delta I_D/I_D)^2 + (\delta I_V/I_V)^2}} = 0.5 \ln (I_V/I_D).
$$

This relationship shows that $u_D$ of the film with an intensity close to that of $I_V$ has a large percentage error. For a relative error of the x-ray intensity of $\sim 20\%$, the error of $u_D$ is about 0.04 nm. Stated differently, a measured displacement of 0–0.04 nm all corresponds to a perfect CoSi$_2$ film. The displacement extracted for the films implanted to $<2 \times 10^{14}$/cm$^2$ after all annealings is $<0.04$ nm (Fig. 4). This suggests that these films recover completely after all annealings. The displacement of the film implanted to $5 \times 10^{13}$/cm$^2$ decreases with increasing annealing temperature, and becomes indistinguishable from that of the as-grown film after 600 °C annealing (Fig. 4).

The heavily damaged CoSi$_2$ films ($\phi > 10^{15}$/cm$^2$) have very different annealing characteristics. Upon thermal annealing, the x-ray peak intensity rises out of the background ($\sim 0.003\%$) and becomes measurable. The peak position is about the same as that of the as-grown film, meaning that the strain of the heavily damaged films does not change after annealing. However, the peak is much broader and weaker than that of the as-grown film, even after 800 °C. This result signifies that the strain in the annealed films is very inhomogeneous, and indicates the presence of extended defects. These results correlate well with those obtained from channeling measurements.
C. Electrical resistivity

We also measured the resistivity in the implanted film with a four-point probe to monitor the evolution of defects. The sheet resistance of the selected samples was also measured by the van der Pauw method, and agreed with that of the four-point probe measurements. Current-voltage measurements show that the vertical resistance across the silicide-silicon interface is always much greater than the sheet resistance of the silicide film for all samples, meaning that the film is practically insulated from the substrate. The resistivity was extracted from the sheet resistance with the film thickness of 50 nm obtained from backscattering measurements. The measured resistivity of the CoSi$_2$ film rises with the dose of ion implantation. This is similar to what is observed for other implanted metallic silicides, but is opposite to what is found for implanted semiconducting silicides (ReSi$_2$ from Ref. 20, CrSi$_2$ from Ref. 21). The resistivity of CoSi$_2$ films increases from $\sim 16 \mu\Omega \text{cm}$ for the as-grown sample to $\sim 300 \mu\Omega \text{cm}$ for the sample implanted to $3 \times 10^{14}$/cm$^2$ and flattens off up to $3 \times 10^{15}$/cm$^2$ (Fig. 5).

The resistivity, $\rho$, of metals can be decomposed into two terms according to Matthiessen’s rule,

$$\rho(T) = \rho_L(T) + \rho_D,$$

where $\rho_L$ is the lattice (Bloch-Gruneisen) contribution, and $\rho_D$ results from carrier scattering by defects at 0 K. Hensel et al.$^9$ measured the resistivity of CoSi$_2$ films bombarded with 2-MeV $^4$He at 4–300 K and found good agreement with Eq. (11). These authors also established that the resistivity of samples implanted to different doses has a similar temperature dependence, meaning that the lattice contribution $\rho_L(T)$ is the same regardless of the damage (so long as the film is not amorphized to cause localization of carriers). The difference of $\rho$ between the bombarded and the as-grown films, $\Delta \rho$, hence equals $\Delta \rho_D$, the resistivity contribution from radiation-induced defects. In an as-grown CoSi$_2$ film, $\rho_D$ is $\sim 2 \mu\Omega \text{cm}$.$^9$ Hensel et al.$^9$ discovered that $\Delta \rho_D$ rises approximately linearly with dose till $\sim 100 \mu\Omega \text{cm}$ and then flattens off with further dose increase. The resistivity jumps abruptly to $\sim 1000 \mu\Omega \text{cm}$ once the CoSi$_2$ is amorphized, and remains constant thereafter. This is also the resistivity of amorphized ReSi$_2$ films ($\sim 1200 \mu\Omega \text{cm}$).$^{20}$ The common value of the resistivity of amorphized metallic and semiconducting silicides suggests that the conduction in such an amorphous state is dominated by hopping of localized charge carriers.

We used the van der Pauw method to measure the resistivity of the CoSi$_2$ film for selected samples at 130–300 K. The temperature dependence of the resistivity agrees with the prediction of Eq. (11), and that reported by Hensel.$^9$ We assume in the following that the resistivity difference $\Delta \rho$ between the implanted and the as-grown CoSi$_2$ films equals that induced by the implantation $\Delta \rho_D$ for all our samples (CoSi$_2$ is not amorphized). $\Delta \rho_D$ increases with dose to $\sim 280 \mu\Omega \text{cm}$ at $5 \times 10^{14}$/cm$^2$ and flattens off up to $3 \times 10^{15}$/cm$^2$ (● in Fig. 5). The different resistivity at plateau between our result and that of Hensel et al.$^9$ (280 vs 100 $\mu\Omega \text{cm}$) is probably due to the different implantation conditions (230-keV $^{28}$Si vs 2-MeV $^4$He) that produce different microstructures of defects in the implanted CoSi$_2$ film.

The resistivity from carrier scattering by point-like defects is proportional to the density of scatterers. The initial rise of $\Delta \rho_D$ in lightly damaged films ($\phi < 5 \times 10^{14}$/cm$^2$) suggests that the implantation produces point-like defects which build up with increasing dose.$^9$ This conclusion is very similar to that drawn from x-ray diffraction results (see previous discussion). Figure 6 plots the resistivity caused by defects scattering as a function of the atomic displacement induced by defects in lightly damaged CoSi$_2$ film ($\phi < 5 \times 10^{14}$/cm$^2$, see Figs. 2, 4, and 5). The plot shows a good correlation between these two indicators of the defect concentration in the CoSi$_2$ films. In heavily damaged CoSi$_2$ films, with dose from $10^{15}$/cm$^2$ to $3 \times 10^{15}$/cm$^2$, $\Delta \rho_D$ is roughly a constant ($\sim 280 \mu\Omega \text{cm}$, Fig. 5), while the defect concentration $\rho_D$ extracted from the channeling yields increases from $\sim 27\%$ to $\sim 54\%$ (● in Fig. 2). This sug-

![FIG. 5. Room-temperature resistivity difference between the as-grown CoSi$_2$ film ($\sim 16 \mu\Omega \text{cm}$) and the implanted ones vs the dose, for the as-implanted sample (●), and for those annealed in vacuum for 60 min at 250°C (V), 400°C (O), 600°C (△), 800°C (square).](image1)

![FIG. 6. Resistivity difference vs the static displacement for the lightly damaged CoSi$_2$ films ($\phi < 5 \times 10^{14}$/cm$^2$). The approximately linear relationship indicates a good correlation between the concentration of the carrier scatterers and the structural defects in such films.](image2)
gests that the flat $\Delta \rho_2$ is probably indicative of some agglomeration of defects in heavily damaged films, in agreement with the results from annealing studies by channeling and x-ray diffraction.

The resistivity of the CoSi$_2$ films of all samples decreases drastically after 250 °C (in Fig. 5), and becomes about the same as that of the as-grown sample after 600 °C (in Fig. 5). The decrease of the resistivity of lightly damaged films ($\phi < 5 \times 10^{14}/\text{cm}^2$) after annealing correlates well with the recovery of structural defects probed by channeling and x-ray diffraction measurements. For highly damaged films ($\phi > 10^{15}/\text{cm}^2$), the structural recovery after annealing is incomplete, with a channeling yield of $\sim 10\%$ (greater than that of the as-implanted films to a dose $< 5 \times 10^{14}/\text{cm}^2$), and with a broad x-ray diffraction peak that suggests the existence of extended defects in these annealed films. Yet the resistivity of such films is about the same as those of the structurally perfect ones, much less than that of the as-implanted films of any dose. Apparently the residual extended defects that still exist in the highly damaged films after annealing are ineffective scattering centers for the carrier transport.

IV. CONCLUSION

In summary, we demonstrate that MeV ion channeling is well suited to characterize heavily damaged CoSi$_2$ films ($e_2 > 1\%$) while x-ray diffraction does best for lightly damaged ones. Resistivity is a relevant indicator of the damage over the entire range. A linear relationship exists between the structural and the electrically active defects in lightly damaged films. Such a relationship is absent in heavily damaged CoSi$_2$, where the resistivity flattens but the defect concentration increases further with dose. Lightly damaged films recover completely upon thermal annealing. Heavily damaged films also recover their original resistivity after annealing, but extended defects (e.g., dislocation loops and microtwins) that are ineffective for carrier scattering subsist.

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