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PRECISION INTERPLANAR SPACING MEASUREMENTS OF BORON DOPED SILICON

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A study of lattice parameters of boron doped silicon (10^{14}-10^{19}atom/cc) grown in <111> and <001> directions by Czochralski technique has been undertaken. Interplanar spacings (d) were measured by pseudo-Kossel technique to a precision up to 0,001% ; different procedures to obtain d and the errors are discussed. It is concluded that the crystallographic planes contract preferentially and the usual study of parameter variation must be made as a function of d. The diffused B particularly contracts the {333} plane and in a more pronunciate way in high concentrations. An orientation dependence of the diffusion during growth was observed.

INTRODUCTION

Silicon crystal-lattice distortion, resulting from impurity diffusion, has been widely studied due to interest in solid-state devices. Pearson & Bardeen (1949) studying the electrical properties of undoped and boron doped silicon verifyed that the presence of substitucional B in the silicon lattice as acceptor impurity produces a lattice contraction, originating elastic deformations. Horn (1955) arrived to the same conclusion employing density measurements in addition to lattice parameters, electrical and chemical analysis. The B diffusion (5×10^{20}at.cm^{-3}) in silicon wafers reduces the lattice parameter due to different values of the ionic radius of Si atoms (1.17Å) and substitucional B atoms (0.98Å) (Queisser, 1961; Carruthers, Hoffman & Ashner, 1963); this reduction introduces uniform deformation and for adequately large values it favours the formation of dislocations, regularly distributed in the slip planes. Cohen (1967) considered that discrepancies found in the contraction coefficient of silicon lattice are due to the non-consideration of deformation produced by diffusion process and the complexes formation; the generation of vacancies in excess would be one of the factors contributing to make the diffusion a function of the temperature (Jain & Overstraeten, 1973). Fukuhara & Takano (1977) employing X-ray techniques confirmed part of the results of Queisser (1961) verifying the proportionality between the elastic deformation and the B concentration; after a critical value of deformation, defects like dislocations appear in order to relax that deformation and proportionality is no longer valid.

The B diffusion process is certainly related to the so called microdefects in Si. Despite intensive efforts, basic questions about point defects in Si are still unanswered or hotly disputed, with the consequence that all the important diffusion

mechanisms of impurities are still not known with certainty. (Föll, Gösele & Kolbesen, 1981). Besides that, some crystallographic aspects of the B difussion process are not well known.

In the previous papers about the silicon crystal-lattice distortion, resulting from boron diffusion, a uniform lattice contraction was assumed since it is usually expressed in terms of a variation Δa of the cubic cell parameter a. This paper presents an accurate study of the crystalline parameter variation of boron doped silicon single crystals grown by Czochralski method and doped by diffusion during growth. The orientation dependence of diffusion of B in silicon slices has been pointed out (Allen & Anand, 1971, among others) but a possible relation among the crystal growth direction, diffusion process and alteration of crystallographic parameters has not been established. This paper also presents a correlation between growth direction (<100> and <111>) and interplanar spacing of the silicon doped lattice during growth.

It was employed the back-reflection pseudo-Kossel technique that can determine lattice parameters to within 0.002% (Berg & Hall, 1974), having the advantage of allowing the separate determination of interplanar spacings from planes of the same family, providing a technique for investigating anisotropy (Newman & Shrier, 1970).

Several methods of measuring interplanar spacings with precision, using the back-reflection X-ray divergent beam technique, have been suggested (for a review, see Newman & Shrier, 1970). The foremost difficulty with this technique is that the pseudo-Kossel lines are complicated fourth degree equations and the assumption that they are conics would lead to significant error (Newman, 1970). The incomplete formation of the curves, due to some sub-structure in the specimen, constitutes another limitation for several methods.

In order to overcome these difficulties Schneider & Weik (1967) have suggested geometrical methods, utilizing double exposures, which are conveniently applicable to the study of anisotropic distortions in single crystals. A discussion of the geometrical limitation of this method was presented by Newman & Shrier (1970) who suggested another method utilizing the coordinates of general points of the lines rather than their special geometrical properties. A double exposure of the film is also employed by Newman & Shrier but the corresponding points are established by utilizing discontinuities produced by fine shadows of a wire grid on the pseudo-Kossel curve; from the measurements of the shadow displacements due to film translation the direction cosines of a particular diffracted beam can be determined.

Aristov, Shekhtman & Shmytko (1974) employed an analogous method utilizing a screen of copper wires ($\emptyset \sim 1$mm) in the form of a radial net to create sharp breaks on the diffraction lines; interplanar spacing measurements have shown that the method guarantees an error $\Delta d/d$ of less than 0.0005 . Employing basically the same experimental procedure, another approach to this technique was given by Aristov & Shmytko (1978); they considered that the precision of lattice parameter measurements can be increased if the calculation of the Bragg angle is obtained by combining the approach based on the direction cosines of three arbitrary diffracted surfaces and the employed in the Debye-Scherrer method; this procedure allows high accuracy ($\Delta d/d = 3 \times 10^{-5}$), but it is comparatively more labour consuming.

Aristov, Shekhtman & Shmytko (1974) have proposed that the precision in the determination of the crystallographic parameters by their method can be improved if necessary by increasing the number of exposures, the number of breaks on the lines, the Δz displacement of the cassette as well as by a special mathematical treatment. The accuracy constraints of this technique are associated with the

comparatively high error in determining each of the coordinates of the gap points (Aristov & Shmytko, 1978).

This paper employs the method suggested by Newman & Shrier (1970) and Aristov, Shekhtman & Schmytko (1974) improving some experimental procedures such as employing a screen of very thin tungsten wires to create the breaks on the diffraction lines (Koishi & Gilles, 1979). It is also done a brief discussion about some procedures to obtain the interplanar spacings as well as the correspondent errors.

EXPERIMENTAL

Silicon single crystals were grown by Czochralski method and doped by diffusion process during the growth. The doping was carried out with acceptor type impurity (boron) at concentrations in the solid up to 10^{19} atoms/cc . The silicon wafers have 4 cm in diameter, nominally 200 μm thick with the surface crystallographic plane perpendicular to the growth direction (<111> and <001>). In order to remove the surface stress introduced by the cut process, the wafers were polished by usual chemical-mechanical process. The silicon wafers of normal semiconductor grade specifications belong to batches usually utilized in microcircuit devices in the Microelectronic Laboratory of the Politechnic School, S. Paulo University.

The concentrations of electrically-active dopants were determined with the four point probe technique; this measurement allows also the verification of the resistive homogeneity of the samples within 0.5%. Thickness homogeneity was verified with a precision of 0.1 μm by a capacitive process. The radius of curvature was determined from a light interferometer and in some samples also by a transmission topographic camera (Lang-type). The profile analysis

of the Bragg reflection in a X-ray double diffractometer was employed to assure the crystalline homogeneity (Brito Filho, 1981; Pimentel & Brito Filho, 1982). The X-ray transmission topography has shown the absence of dislocations and surface imperfections due to cut and polishment processes; the X-ray topographic section has shown the typical fringes of a perfect crystal.

The samples with boron concentration higher than 10^{18} atoms/cc , despite their thin thickness, were completly (100%) absorvent to infrared radiation, making impossible any infrared absorption spectroscopic analysis. For the samples with lower B concentration this analysis was performed in a differential form, employing a standard floating zoned silicon crystal from the General Diode Corp. (200 Ω.cm resistivity) with low carbon and oxigen concentration; the differential curves obtained did not show oxigen or carbon absorption lines. Considering the growth conditions, the electrical performance and the general crystallographic characteristics it is possible to admit for the samples the usual values for carbon ($\sim 10^{17}$ atoms/cc) and oxigen ($\sim 10^{18}$ atoms/cc) concentration for Czochralski-silicon (Liaw, 1979). The characteristics of the silicon samples (growth orientation, electrical resistivity and boron concentration) are shown in Table I, together with the experimental results.

The pseudo-Kossel back-reflection patterns were obtained with a Rigaku "Microflex" apparatus of the Physics Department, ITA, employing CuKα radiation (40kV , 65μA) with a focal spot of 60 μm ; a screen of tungsten wires ($\phi \sim 0.2$ mm) in the form of a quadratic net of 20 cm x 25 cm between the crystal and the film produced the "shadows" on the patterns. The focus-screen distance was 55 mm and 75 mm respectively in the first and second exposures. It is necessary to avoid the contribution of the crystal edge to the patterns since the polishing process may change the edge thickness, with consequent

sample	growth direction	resistivity (Ω . cm)	concentration (atom/cc)	mean interplanar spacing (Å) {533}	{620}	{531}	{440}	{333}
1	<100>	19	$7,0 \times 10^{14}$	0,82854 0,00016	0,85829 0,00011	0,91894 0,00046		
2	<111>	9	$1,5 \times 10^{15}$	0,82743 0,00009	0,85862 0,00030	0,91757 0,00059	0,95999 0,00053	1,04442 0,00020
3	<100>	0,9 - 1,2	$1,6 \times 10^{16}$	0,82825 0,00012	0,85840 0,00022	0,91883 0,00042		
4	<111>	0,015	$7,0 \times 10^{18}$	0,82750 0,00038	0,85859 0,00027	0,91722 0,00042		1,04405 0,00017
5	<100>	0,0108	$9,6 \times 10^{18}$	0,82810 0,00011	0,85808 0,00038	0,91751 0,00023		
6	<111>	0,0058	$2,0 \times 10^{19}$	0,82726 0,00033	0,85866 0,00060	0,91640 0,00087	0,9596 0,0012	1,04157 0,00022

TABLE 1 - General characteristics of boron doped silicon samples, mean interplanar spacings for {hkℓ} planes and the corresponding standard deviation of the mean value (68% confidence level).

variation of the sample curvature. On the other hand, only for sample-focus distances > 3,0mm it is possible to obtain the pseudo-Kossel lines corresponding to {533} and {444} of crystals with growth direction respectively <111> and <001>. These facts have limited the sample-focus distance to 3,0 - 3,5 mm. The samples were mounted vertically in a goniometer and set parallel to the film cassette. Sakura and Kodak films with one emultion removed were used.

Several factors contribute to the precision of the measurements and carefull calculations or estimates were made for each source of error. The temperature of the sample during exposure can be taken to be within $\pm 0.2^{o}C$ and the determined interplanar spacing values correspond to $21^{o}C$. For the silicon, a temperature change of $0.1^{o}C$, near $25^{o}C$ and $2\theta_B = 157^{o}$ (Hubbard, Swanson & Mauer, 1975), alters this $d_{2\theta}$ interplanar value in $\sim 4 \times 10^{-5}$%. A variation of $\pm 2^{o}C$ changes both extreme measured interplanar spacings in about 0,001% or, in other terms, changes d_{004} and d_{444} respectively by $\pm 5 \times 10^{-6}$ Å and 5×10^{-7} Å. Hence, changes in temperature several times the controlled tolerance would not affect the calculated interplanar spacings. The error due to sample curvature (Berg & Hall, 1974) is less than 10^{-6} Å since the curvature radius is bigger than 200 m, except for sample 2 (curvature radius $\sim$ 8m) of which the estimated error is less than 3×10^{-5} Å. The calculated precision in the displacement z of the cassette and the estimated buckling of the film in the cassette show that both may introduce an error in $d_{hk\ell}$ less than 0.05%. The shrinkage of the photographic material was calculated and the error introduced by the development process in $d_{hk\ell}$ is about 0.001%. Measurements on the film were made with an Enraf-Nonius microdensitometer fitted with a 3 x telescope; this coordinate measuring constitutes the main source of error and is due to the absolute uncertainty in the reading (± 0.005mm), the observator,

the line width and the edges of breaks, despite their sharp feature. In order to minimize this error, for each particular (hkℓ) plane the corrdinates (x,y) were measured from seven to twelve pairs of discontinuities on the (hkℓ) diffraction lines.

CONSIDERATIONS ABOUT THE PSEUDO-KOSSEL METHOD

The pseudo-Kossel lines were indexed by comparison with a theoretical pattern obtained by a computer program (Koishi & Gilles, 1979) that takes into account the experimental conditions and gives reflection conics on the crystal surface. Table 2 shows the plane family {hkℓ} and reflections (hkℓ) appearing in the pseudo-Kossel patterns in the particular experimental conditions for samples with <111> and <001> growth directions.

A computer program has calculated the direction cosines of the diffracted rays defined for each pair of coordinates (x,y) and the displacement z of the film, providing N directions (N=7 to 12 in this paper); the direction cosines of the normal to a (hkℓ) plane was obtained from the system of equations (Newman & Shrier, 1970):

$$\vec{s}_i \cdot \vec{m} = \sin\theta_i \qquad (i = 1,2...N)$$
$$|\vec{m}| = 1 \qquad (1)$$

where $\vec{s}_i$ represents the direction cosine of the diffracted ray (DCD), the components of $\vec{m}$ represent the direction cosines of a solution for the unit vector normal to a (hkℓ) plane (DCN) and θ the Bragg angle. Two different procedures (I and II) were adopted in order to obtain DCN, the interplanar spacings (hkℓ) and {hkℓ} as well as the corresponding errors.

{hkℓ} plane	(hkℓ) plane <111>	(hkℓ) plane <001>
444		$(\bar{4}44)(4\bar{4}4)$
533	(353)(335)(533)	$(\bar{3}35)(3\bar{3}5)(33\bar{5})$
620	(260)(062)(026)(206)(602)(620)	$(\bar{2}06)(0\bar{2}6)(20\bar{6})(026)$
531	(351)(153)(135)(315)(513)(531)	$(\bar{3}15)(3\bar{1}5)(\bar{1}35)(1\bar{3}5)(31\bar{5})(\bar{1}35)$
440	(440)(044)(404)	
511		$(\bar{1}15)(1\bar{1}5)(11\bar{5})$
333	(333)	
242	(242)(224)(422)	
004		(004)

TABLE 2 - (hkℓ) reflections appearing in pseudo-Kossel patterns for samples with <111> and <001> growth directions parallel to the incident beam.

By procedure I one solution was determined for the DCN for each three triplets of DCD, which means to obtain $N(N-1)(N-2).16 = N'$ solutions or N' values for the DCN and N' values for $\sin\theta$; employing the Bragg law and the λ value, N' values for $d_{(hk\ell)}$ were obtained (Koishi & Gilles, 1979). In this present paper selected data reducing the N' values to $n \leq N'$ were employed in order to avoid the introduction of rough errors in the mean value $\overline{d}_{(hk\ell)}$. The mean interplanar spacing of a $(hk\ell)$ plane $(\overline{d}_i \equiv \overline{d}_{(hk\ell)})$ was given by the arithmetic average for the n values; σ_i is the standard deviation of the distribution.

For a $\{hk\ell\}$ plane family,

$$\overline{d} = \frac{\sum_{i=1}^{r} \omega_i \overline{d}_i}{\sum_{i=1}^{r} \omega_i} = \frac{\sum_{i=1}^{r} \frac{n_i \overline{d}_i}{\sigma_i^2}}{\sum_{i=1}^{r} \frac{n_i}{\sigma_i^2}} \qquad (2)$$

where r corresponds to the number of $(hk\ell)$ planes of a same $\{hk\ell\}$ family, and $\omega_i = n_i/\sigma_i^2$. The standard deviation of this global mean value is given by

$$\sigma_{\overline{d}}^2 = \frac{\sum_{i=1}^{r} \sum_{j=1}^{n_i} \frac{(d_{ij} - \overline{d})^2}{\sigma_i^2}}{(n-1) \sum_{i=1}^{r} \sum_{j=1}^{n_i} \frac{1}{\sigma_i^2}}$$

Employing algebraic artifices and definitions of mean values and deviation, it is possible to simplify and rewrite this last expression:

$$\sigma_{\overline{d}}^2 = \frac{n - r + \sum_{i=1}^{r} \frac{n_i}{\sigma_i^2} (\overline{d} - \overline{d}_i)^2}{(n-1) \sum_{i=1}^{r} \frac{n_i}{\sigma_i^2}} \qquad (3)$$

Procedure II followed the approach suggested by Newman & Shrier (1970); the equation (1) was rewritten in a more suitable form:

$$s_{zi} = a\, s_{x_i} + b\, s_{y_i} + c$$

where

$$a = -\frac{n_x}{n_z} \quad , \quad b = -\frac{n_y}{n_z} \quad \text{and} \quad c = \frac{\operatorname{sen}\theta_i}{n_z}$$

and $i = 1,\ldots,N$.

In this paper, in order to determine the parameters a,b,c of the system of N equations the REGRE/IBM program was used employing multiple linear regretion (Ostle, 1954). The mean interplanar spacing of a $(hk\ell)$ plane is given by

$$\overline{d}_i = \frac{\lambda}{2} \frac{\sqrt{1 + a^2 + b^2}}{|c|}$$

$(\lambda(Cu K\alpha_1) = 1.5405981\ \text{Å})$.

For a $\{hk\ell\}$ plane family, the average interplanar spacing $\overline{d}$ is given by the arithmetic average of the r values of $\overline{d}_i$

$$\overline{d} = \frac{\sum_{i=1}^{r} \overline{d}_i}{r} \qquad (r = 3 \text{ to } 8) \qquad (4)$$

and the corresponding standard deviation:

$$\sigma^2_{\bar{d}} = \frac{\sum (\bar{d}_i - \bar{d})^2}{r - 1} \qquad (5)$$

The values of $\bar{d}$, $\sigma^2_{\bar{d}}$ (and the corresponding standard deviation of the mean value, ε) obtained from procedures I (expressions 2, 3) and II (expressions 4,5) are shown in Table 3 for some planes $\{hk\ell\}$.

$\{hk\ell\}$	procedure	$\bar{d}$	$\sigma_{\bar{d}}$	$\varepsilon_{\bar{d}}$	n
531	I	0.91758	0.00397	0.00018	487
	II	0.91751	0.00065	0.00023	8
620	I	0.85803	0.00135	0.00010	198
	II	0.85808	0.00064	0.00038	4
335	I	0.82800	0.00162	0.00012	182
	II	0.82810	0.00017	0.00011	4

TABLE 3 - Values of mean interplanar spacing ($\bar{d}$), standard deviation ($\sigma_{\bar{d}}$), standard deviation of the mean value ($\varepsilon_{\bar{d}}$) and degrees of freedom (n) for $\{hk\ell\}$ planes obtained from procedures I and II (sample 5).

The relative discrepancy between the $\bar{d}$ values obtained from procedures I and II is in general less than 0.01% and always whithin the standard deviation of the mean value. The largest difference between these procedures occurs in the $\sigma_{\bar{d}}$ values since procedure I leads to a set of $d_{(hk\ell)}$ values much more dispersive than procedure II, despite the mean value being practically the same, as well as their standard error (perhaps a little subestimated in procedure I).

Expressions (1) can be thought as equations of a plane where $\sin\theta_i$ is the origin-plane distance and the components of $\vec{s}_i$ are coordinates of a point belonging to that plane. Procedure I determines a plane for each triple of points and so the corresponding $d_{(hk\ell)}$ value; this means that not all these $d_{hk\ell}$ values are independent. By this procedure, it was attributed equal weight for sets of very close and very distant points. This implies a not very well defined average and an overestimated dispersion among the $d_{(hk\ell)}$ values; therefore it is impossible to employ the $\sigma_{\bar{d}}$ values in order to estimate the dispersion among the r values of $\bar{d}_{(hk\ell)}$ planes which would give an idea about the anisotropic alterations in planes belonging to the same family. Procedure II adjusts the best plane containing the N points and the distances between the points are taken as weight. It was considered that this last procedure leads to a best determination of $\bar{d}$ values with a improved precision ($\sigma_{\bar{d}}$ low) and it was utilized in this paper.

RESULTS AND DISCUSSION

A typical back-reflection pseudo-Kossel pattern is shown in Figure 1. The pseudo-Kossel lines of all the patterns did not show displacements, breaks or preferential enlargement; dislocations were not detected. This result together with the other characterizations allows to consider the samples as "free dislocations".

Table 1 shows the values of mean interplanar spacing and the corresponding standard error for the most significant planes. It is worthwhile to note that although it is possible to obtain in the pseudo-Kossel patterns the $(hk\ell)$ reflections presented in Table 2 some of them are not adequate for precise measurements: low order reflections are characterized by large errors; reflection lines (444)

cover a very small area and are not always completely available in the film.

The usual procedure to study the lattice parameter variation in a cubic system investigates the cell parameter a. The precision determination of this a parameter by X-ray divergent beam method was carried out by adopting the sequence of steps given by Ellis et al (1964). For this determination the lattice parameter a' was computed using the relation $a' = d(h^2+k^2+\ell^2)^{1/2}$ and to each a' of a $\{hk\ell\}$ family it was obtained a value of the Nelson-Riley function; employing a method of least square, the lattice parameter a_o was obtained by extrapolation of the Nelson-Riley plot of the weighted $(1/(\sigma_{\overline{d}})^2)$ a' values. For samples with high B concentration the analysis of this graph has shown a dispersion from that linear function, higher than the experimental error. Besides that, it was observed that the more dispersive a' values have presented high $\sigma_{\overline{d}}$ values meaning that practically they did not contribute to the a_o value extrapolated. If it is taken into account the physical significance of $\sigma_{\overline{d}}$, intimately related to the residual strain distribution in the crystal (Ellis et al 1964) it can be seen that the planes $\{hk\ell\}$ affected by the boron diffusion process are those that less contribute to an estimate of the lattice parameter alteration due to boron diffusion. Of course, that procedure would have been valid if the B diffusion process should affect equivalently all the planes, what means a isotropic contraction. Therefore it is considered that in the present case a study about the influence of concentration must be made considering alterations in the lattice interplanar spacings.

Figure 2 shows the graphic of mean interplanar spacings of several planes $\{hk\ell\}$ related to boron concentration values; the plotted values are identificated according to the growth directions of the samples. The interplanar spacings for a high-purity perfect

silicon measured by the same X-ray method (Aristov & Shmytko, 1978) are also shown in the graphs in order to compare the lattice interplanar spacings of undoped and B doped silicon single crystal.

The diverse behaviour of the distinct planes, suffering preferential contractions or little affected by the doping is evident. The actual occurrence of this particular behavior did not allow the usual analysis of the lattice contraction coefficient to be carried out as a function of doping concentration taking into account just an average of the cell parameter a_o. This is particularly serious if the boron concentration is 10^{18} - 10^{19} atoms/cc.

As a general tendency a decrease of the $\overline{d}$ values with the increase of B concentration can be seen and in a more pronunciate form for 10^{18} - 10^{19} atoms/cc, meaning that in general the interplanar spacings tend to contract. Nevertheless the $\{hk\ell\}$ planes have suffered different contractions, being the {333} the most affected. The {620} plane did not suffer marked alterations.

An interesting aspect has been the systematic behavior of samples depending on the growth direction. It can be noted that interplanar spacings of {533} and {531} planes from <111> are smaller than those for the <001> samples; in a not well definitive form the reverse seems to occur for the {620} planes. This indicates a orientation dependence of the diffusion of boron in silicon during growth, analogous to observation in silicon slices where boron depositions were carried out (Allen & Anand, 1971); (Allen, 1973). For both growth directions each particular reflection shows similar dependence of the interplanar spacing with B concentration.

The causes of that contraction should not be immediately attributed to the presence of substitutional boron in the silicon lattice. The two major non-intentionally doped impurities in silicon crystals, oxygen and carbon, may also contribute to alter the lattice parameters. The oxygen is incorporated into the silicon lattice as a

solid solution occupying interstitial sites, causing a lattice expansion; on the other hand the carbon occupies substitutional sites, contracting the Si lattice, but it may also be located intersticially (for concentration $\sim 10^{18}$ cm^3), expanding the lattice (Liaw, 1980). It is possible that these impurities may contribute for some non uniform behavior of the d parameter in function of B concentration. Nevetheless, based on the previous considerations about the presence of C and O , it is possible to admit the marked contraction in {333} plane for concentrations $\sim 10^{18} - 10^{19}$ atoms/cc as due to the presence of the doping. This could be understood if it were admited the substitutional boron located preferentially in the planes {111} but it is desirable a more precise analysis in order to exactly locate the substitutional boron in the crystal cell. However the possibility of a preferential distribution of microdefects particularly in heavy doped silicon samples is not discarded (Brito Filho, 1981).

CONCLUSIONS

It is concluded that the boron doping process in silicon during the Czochralski growth introduces preferential alterations in the Si lattice; the usual study of parameter variation must be done as a function of interplanar spacings of the lattice since the diffusion process affects differently the crystallographic planes. The presence of boron in the Si lattice produces a pronunciate way for concentrations up to 10^{18} atom/cc . This reinforces the hypothesis of the presence of substitutional B in the Si lattice. For identical crystallographic planes the interplanar spacings of grown samples in different orientations suffer distinct alterations with the B concentration. Hence occurs a orientation dependence of the diffusion of B in Si during the Czochralski growth.

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FIGURE CAPTIONS

FIG. 1 - A typical back-reflection pseudo-Kossel pattern of B doped Si free dislocation (part of the pattern).

FIG. 2 - Variation of mean interplanar spacings d of $\{hk\ell\}$ planes with the boron concentration C of silicon samples grown in <111> (⧫) and <001> (⊠) directions (⊥ values from Aristov & Shmytko, 1978).

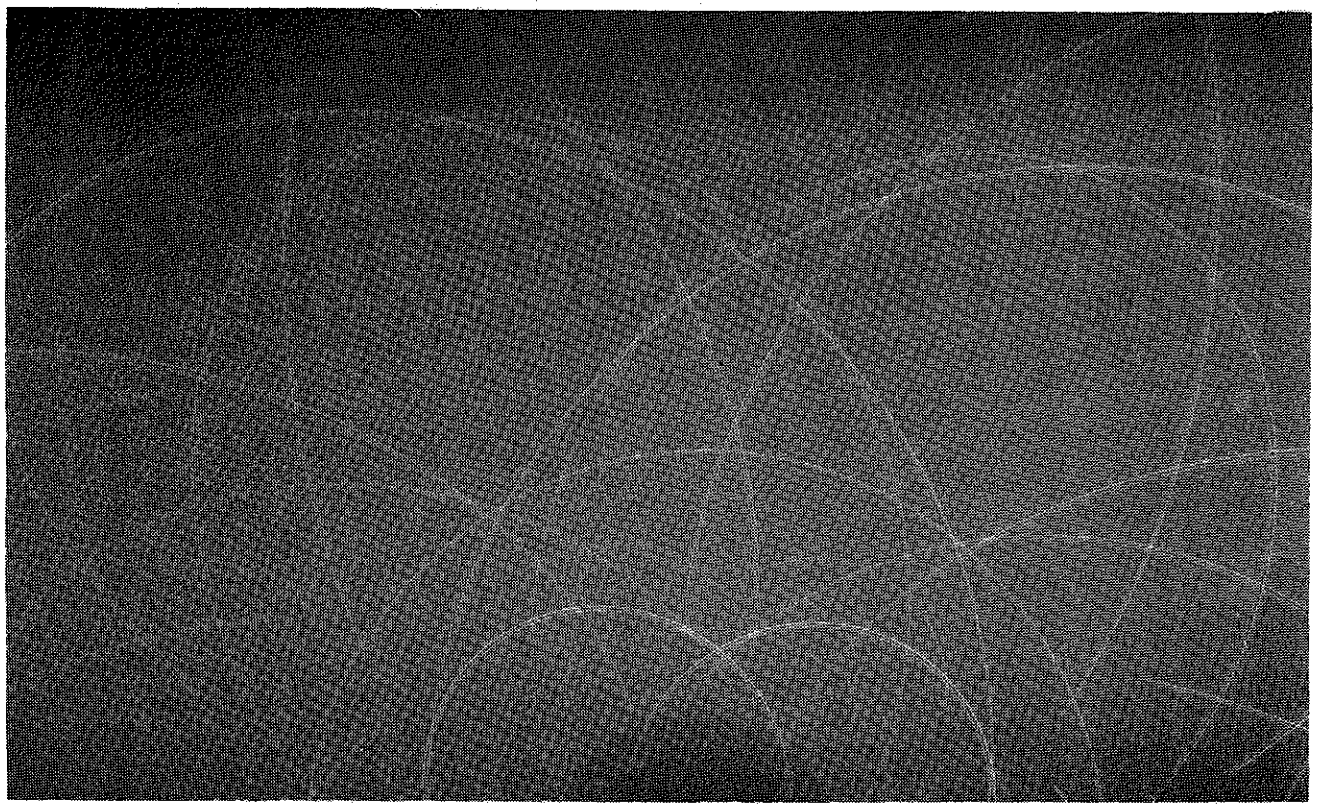

FIGURE 1

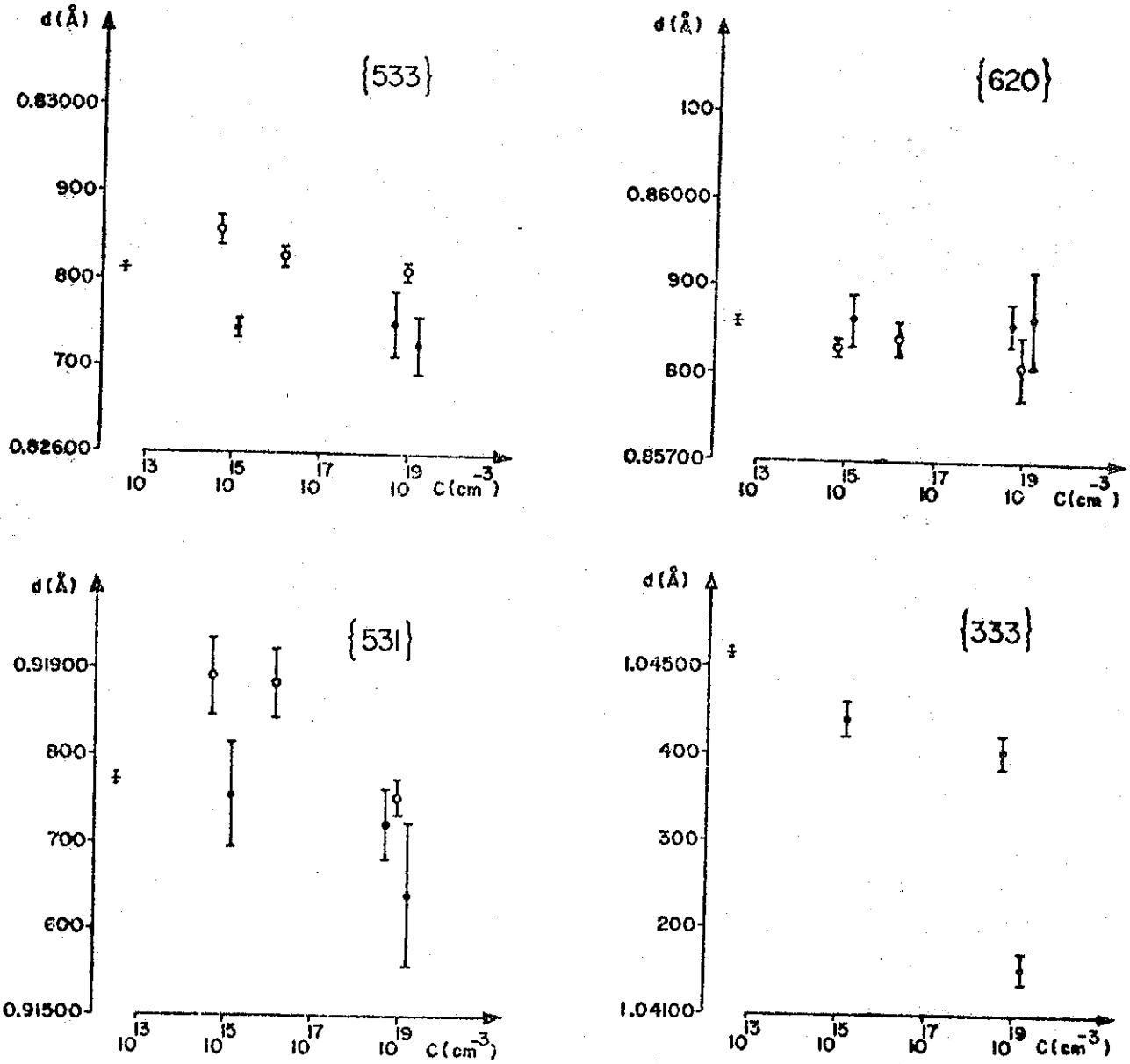
d(Å)
{533}
0.83000
900
800
700
0.82600
10^13 10^15 10^17 10^19 C(cm^-3)
d(Å)
{620}
100
0.86000
900
800
0.85700
10^13 10^15 10^17 10^19 C(cm^-3)
d(Å)
{531}
0.91900
800
700
600
0.91500
10^13 10^15 10^17 10^19 C(cm^-3)
d(Å)
{333}
1.04500
400
300
200
1.04100
10^13 10^15 10^17 10^19 C(cm^-3)

FIGURE 2