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**Variability and character of hollow macro-defects in silicon carbide wafers:
evidence from scanning electron and polarization microscopy**

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Keywords: silicon carbide; macro-defect; polygonized void; micropipe; polarization microscopy

Abstract: Polarization microscopy is a suitable tool for studying strain in appropriately cut SiC single-crystals. Both the outline shape examined by electron microscopy and the induced interference pattern observed by polarization microscopy were used to study the variation and character of macro-defects present in SiC wafers. While voids are usually in a relaxed state, hollow-core dislocations are characterised by large interference halos up to ~ 100 µm in diameter. Conoscopy, i.e. evaluation of the interference pattern created by inserting an Amici Bertrand lens, is suitable to examine these optical phenomena in more detail and to gain additional knowledge on the inclination of cut wafers towards the c-axis. The discrepancy between simulated interference patterns for (0001)-SiC and the observed ones strongly indicates that pipes are no pure screw dislocations as commonly thought, but must have an edge component.

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1. Introduction

Silicon carbide (SiC), long known and used as a ceramic for abrasive or high-temperature applications (Ref. [1, 2]), has become an attractive semiconductor material, because its large band gap width (2.4 - 3.3 eV) and its high critical field strength (3.108 V/m) along with high thermal conductivity and temperature stability allows the use as a high-power and high-temperature device.[3] Its among large band-gap semiconductors unique property to form a solid silica scale during thermal oxidation enables the fabrication of SiC/SiO₂ devices as metal-oxide-semiconductor field effect transistors (MOSFETs) - quite similar to Si/SiO₂. [4] Hence it is not surprising that SiC merits the rank of both, being the single most important non-oxide ceramic material and the most promising semi-conductor material to replace silicon based devices.

However, the most challenging obstacle to overcome for a large-scale application of SiC-based devices is to avoid structural defects during single-crystal growth. The structural diversity of SiC with its more than 200 different modifications (polytypes; Ref. [5]) along with defects formed during the growth process limit both possible crystal sizes and single-crystal quality. Silicon carbide single-crystals grown by physical vapour transport (PVT) in the modified Lely method [6] - today the most widely used method to obtain large-scale SiC wafers - are highly susceptible to contain defects such as tube-like defects (micropipe) or polygonized, tabular / platelet-shaped voids.

2. Macro-defects in silicon carbide

2.1 Pipes

The so called “pipes” this paper is concerned with are hollow, tube-like structures with diameters from below 0.1 up to more than 5 µm. These defects are believed to be hollow-core screw dislocations with the magnitude of a Burger’s vector in case of small-period polytypes being even larger than one unit cell.[7] Sub-micrometer micropipes may be referred to as

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nano- or micropipes *sensu stricto* and those with larger diameters, as “macropipes” [8]
Usually, micro- and macropipes are orientated fairly well along the c-axis in hexagonal
silicon carbide crystals. [9] However, there are macropipes, which show a serpentine-like
structure with significant deviation from the direction of the c-axis. [10]
Today by critical control of process stability and growth in off-c-axis orientations defect
densities have been lowered significantly. [11] Nonetheless the nature and origin of pipes
remain a matter of dispute and mosaicity-, overgrowth- and void-movement- models are just a
few of the explanations proposed.

2.2 Voids

Polygonized voids are also very prominent defects present in PVT-grown SiC. In hexagonal
SiC, these hollow structures have a hexagonal parallelepiped outline and their edge faces lie
perpendicular to the (0001) face. [12, 13] In contrast to pipes, these defects can reach
diameters larger than 500 µm and are therefore even macroscopically visible as “negative
crystals” or “hollow crystals” with reflecting, inner faces. Typical aspect ratios between
height (h, parallel [0001]) and width (w, perpendicular [0001]) are $h/w \geq 5$. [10] These voids
seem to be able to move along the growth direction. [10, 12, 14] and hence may result from
imperfections at the seed interface. Movement of voids requires a certain diameter (Ref. [15])
and can lead to (off-axis) pipe formation. [12] The movement of such defects might be
explained according to Hofmann et al. (Ref. [10]) by local differences in the heat transfer
distribution and different kinetics at the sublimation (bottom) and growth face (ceiling). The
fact that critical control of the seed interface is capable to reduce the defect density of voids
significantly [16] points in this direction.

2.3 Voids terminating pipes

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Voids often terminate pipes. The model by Kuhr et al. [12] argues that pipes, which are not induced during initial growth of the seeding crystal must be caused by condensation of smaller screw-dislocations. In direction of the growing surface Kuhr et al. [12] noticed dislocations, which were not connected to the initial seed. Instead, pipes originated from the voids' rims and propagated towards the seed. Moreover, evaporation steps on the voids' top and growth steps on the bottom were detected, which was taken as evidence for a movement of voids by a local vapour transport process along the thermal gradient, perpendicular to the growth face.

Dislocations, however, can neither end within a compact crystal nor within a cavity inside a wafer and the absolute value of all Burgers vectors on top of a void must be the same as on the bottom side. Condensation of individual screw-dislocations at the void's edge forming open-core dislocations is due to the convex form of void's bottom part (cf. Fig. 1). A recrystallization within a small cavity, with opposing faces that are only a few micrometers apart from each other, should proceed towards the edges, until a critical value in dislocation density is reached. The low temperature gradient and the lower oversaturation in the cavities explains the low level of dislocations because of the low nucleation rate according to Glass et al. (Ref. [17, 18]) and Ma (Ref. [19]). Also, a growing face with a line vector parallel will push the dislocation along the growth front: comprising the dislocation would induce an energetically disadvantageous stress field.

Once the growing area on the bottom part of the void has reached the voids edges, the dislocations collected will be condensed into clusters whose stress field makes any further approach of the growth isle impossible and an open-core dislocation according to Frank (Ref. [20, 21]) must remain.

2.4 Defect characterization

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To detect and analyze structural defects in silicon carbide, defect selective etching (e.g. using molten KOH) can be used. Subsequently, other characterization methods can be applied - among them quite simple ones like visual inspection using optical microscopy (e.g. Nomarski Differential Interference Contrast NDIC). The basic principle behind the concept of preferential etching is that areas with high internal strains tend to be etched more rapidly, thus creating characteristic etch pits on the surface.[17, 18] For higher resolution, Scanning or Transmission Electron Microscopy (SEM or TEM, respectively) or Atomic Force Microscopy (AFM) can be applied. More information, including crystallographic misalignment, can be obtained by using Synchrotron White Beam X-ray Topography (SWBXT) and High-Resolution X-Ray Diffraction (HRXRD).[22-25]

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The latter methods are certainly more elaborate than simple optical microscopy. However, using transmitted light microscopy, small macropipes and especially micropipes lying along the c-axis cannot be seen on single-crystals cut perpendicular to c and only tilted pipes will appear as one focuses through the sample. For determining defect densities, a quick and reliable method needs to be applied. As shown in the study described in this paper, simple polarization microscopy is truly capable of coping with this task and conoscopic imaging furthermore reveals the wafer's inclination towards the c-axis. Polarization microscopy to describe stress-related interference pattern due to inclusions or dislocations has successfully been used for more than 50 years.[26-28]

3. Experimental method

For electron microscopy a LEO VP 1450 instrument (Carl Zeiss AG, Oberkochen, Germany) was applied operating at 15 kV acceleration current. An Olympus BH2-UMA Normarsky phase contrast polarisation microscope (Olympus Europa Holding GmbH, Hamburg, Germany) was used for polarization microscopy. Simulation of stress field was carried out using Matlab 7 (The MathWorks, Inc, Boston, U.S.A.).

4. Results and discussion

4.1 Morphology of macro-defects

Macro-defects in single-crystal silicon carbide exhibit a large range of structural variation. It is a common mistake to assume macropipes to behave like micropipes and postulate an even, well-shaped habitus. This is not true, even for micropipes. Fig. 2 depicts some examples of various pipe outlines. Only high-quality wafers tend to show less morphological variety and exhibit rather uniformly shaped, often polygonized (hexagonal) hollow-core screw-dislocations (Fig. 2h).

The direction of the c-axis is often the growth direction, and most pipes are oriented along it with little aberration.^[29] The deviation seems to be related to faceted growth: pipes on areas, where faceted growth occurs, are almost ideal orientated along the c-axis, pipes grown on non-faceted areas can occur with a certain angle in respect to the c-axis.^[9]

However, also randomly orientated, short range pipes have been observed.^[30] SiC single-crystals grown perpendicular to the c-axis indeed did not show any pipe at all, but were characterized by a high level of stacking faults.^[11, 31] In our samples we found pipes following the outer curvature of the sliced ingot rather than the c-axis (Fig. 3). Their lateral expansion is limited ($\ll 100 \mu\text{m}$), but their vertical length often comprises the whole height of the single-crystals, which were up to several centimetres thick.

Large, often polygonized voids (“hollow crystals”) can be observed as a typical macro-defect lying in the basal plane. Depending on the SiC polytype, quite different shapes of these voids can be seen. For hexagonal SiC, these voids tend to form rather hexagonal outlines and are therefore often referred to as “hexagonal voids”. However, it is important to note that even in one specific polytype the overall outline of these voids may vary greatly as seen in Fig. 4.

In contrast to pipes voids are vertically quite limited. As shown in Fig. 5a a void with a maximum of $\sim 300 \mu\text{m}$ lateral expansion has a width of $\leq 50 \mu\text{m}$ and only a deeper trench on

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the right. Voids, as discussed earlier, often terminate pipes. Such a typical case is depicted in Fig. 5b and 5c.

4.2 Polarization microscopy

4.2.1 Pipes

As discussed in detail by Kato et al. [32], pipes are well visible under the light microscope in the crossed poles condition. A unique interference pattern is created by the distribution of internal stresses (Fig. 6). These “bright star patterns” will change by rotating the sample and are much larger than the pipes they are connected to. Therefore they indicate very precisely the position even of smaller pipes. As shown in Fig. 7 the visible strain field of an approximately 1 μm large pipe can easily reach 50 μm or more.

The origin of these halo like patterns is strain acting perpendicular to the c-axis, because without it no birefringence at all would be seen on the (0001)-face of silicon carbide, where the c-axis and the optical axis converge. [32]

This behaviour is well known e.g. from natural, deformed quartz grains and is referred to as undulatory extinction due to a locally anomalous biaxial optical character. [33] Using conoscopic microscopy the anomalous biaxial optical character can easily be made visible. In case of (0001)-SiC, normally the bisecting isogyres won’t change when rotating the sample around the c-axis. Strain around pipes causes the isogyres of a strained section to show an anomalous biaxial figure (i.e. a small opening of the central cross = melatope; Fig. 8). Also, as seen in Fig. 8c, inclination to a cut perpendicular to the c-axis can easily be detected by non-centric rotation of the isogyres.

A more detailed study of the outline of the interference pattern around pipes is shown in Fig. 9. The outline of the interference pattern changes when rotating the sample along the c-axis (Fig. 9a). A very simple equation for edge dislocations [26] was found suitable. K in Eq. 1 can be seen as a scaling factor.

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$$r^2 = K^2 \cdot \cos^2(2\phi) \cdot \cos^2(\phi - \beta) \quad \text{Eq. 1}$$

This equation uses r and ϕ to describe any point in polar coordinates and β describes the angle between the polarizer and the slip direction. K is defined as (Eq. 2):

$$K = \frac{\left(\frac{2a \cdot A \cdot \pi \cdot t \cdot C}{\lambda}\right)^2}{T} \quad \text{Eq. 2}$$

Here, a is the amplitude of the incident plane-polarized light, t denotes the plate's thickness, C is the mean strain-optic coefficient, λ describes the wavelength and T denotes the intensity of transmitted light.

The factor A is calculated using Eq. 3, where b is the absolute value of the Burger's vector and ν the Poisson's ratio.

$$A = \frac{b}{4\pi \cdot (1 - \nu)} \quad \text{Eq. 3}$$

Simple polarization microscopy is therefore capable of characterizing strained regions in SiC in general and especially around pipes as long as the stress field is large enough for optical resolution.

Introducing a red-I-plate yields additional information about the direction of strain (cf. Fig. 9b). The presence of a homogenous, superimposing stress field would be indicated by bright and dark branches in the interference pattern (Fig. 9c).

The equations above fit the phenomenological observations perfectly. However, this raises questions about the true nature of open-core dislocations. We will discuss this topic in more detail below. Firstly, we will compare the stress field around (1) pure screw dislocations and around (2) screw dislocations having an edge component. A simulation of the stress field around both dislocation types using the model of Hirth et al. (Ref. [34]) and the elastic constants of 6H-SiC from Kamitani et al. (Ref. [35]) is shown in Fig. 10. It illustrates the lines

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of constant intensity for (a) a pure screw-dislocation (Burgers vector $\vec{b} = \begin{pmatrix} 0 \\ 0 \\ 1 \end{pmatrix}$, line vector

$\vec{l} = \begin{pmatrix} 0 \\ 0 \\ 1 \end{pmatrix}$) and (b) a mixed-character dislocation ($\vec{b} = \begin{pmatrix} 0 \\ 0 \\ 1 \end{pmatrix}$, $\vec{l} = \begin{pmatrix} 1 \\ 1 \\ 0 \end{pmatrix}$)

From Fig. 10 (comparison screw and mixed type dislocation) and Fig. 11 (comparison observation and simulation) and other interference patterns in the literature [26-28, 36-38], it is evident that the interference pattern around hollow-core dislocations (pipes) cannot be explained solely by the presence of a pure screw-dislocation but critically requires the presence of an edge-component.

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4.2.2 Voids

As seen in Fig. 12 there is no significant strain field originating from voids at all. Only in cases in which a micropipe originates from the void, a pipe-related interference pattern appeared. Even in highly strained areas at the edge of SiC single-crystals the voids were free of any stress field induced anomalous birefringence, which brightly illuminated other parts of the bulk material. This is shown in Fig. 12c, where strained areas of SiC are separated by a grain boundary.

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This behaviour is expected, if the local vapour transport process leaves the voids moving during single-crystal growth at the bottom with a very low level of dislocations. However, the trench, located at the voids outer border (cf. Fig. 1), often comprises several screw-dislocations (both closed-core and open-core). These dislocations are the source for the interference pattern around voids.

5. Character of pipes in silicon carbide single-crystals

As discussed by Frank (Eq. 4; Ref. [20, 21]) a high level of mechanically induced strain makes it energetically more advantageous to form hollow-core instead of closed-core dislocations during single-crystal growth, as more elastic stress-related energy can be saved than surface energy must be spent. This explains why these concave surfaces, which usually are readily closed during PVT growth, are still present and form those open-core screw-dislocations. It is important to note that a completely undistorted single-crystal would still be energetically highly favourable.

$$r = \frac{G \cdot b^2}{8 \cdot \pi^2 \cdot \gamma} \quad \text{Eq. 4}$$

Eq. 4 correlates an equilibrium radius of the open-core dislocation r with the shear modulus G , the absolute value of the Burgers vector $\bar{b}$ and the surface energy γ .

In literature there are many contradicting findings on the character of pipes. Heindl et al. suggested a significant edge component in small micropipes, while true micropipes were described as almost ideal screw-dislocation using Frank's equation.[39] From Synchrotron

White-Beam X-ray Topography (SWBXT) and simulations, Huang et al. claimed that the micropipes observed in their study were pure screw-dislocations.[23, 40] but Ohsato et al. critically evaluated the strain-field around micropipes by polarization microscopy and found that it was not compatible with a pure screw-dislocation and needed the introduction of an edge component.[37, 38] Although the work of Ohsato et al. contained some inconsistencies

in describing the interference patterns, our own investigations strongly confirm the presence of an edge component.

AFM and SWBXT work presented by Ma et al. (Ref. [41-43]) supported the interpretation of pipes as pure screw dislocations. They combined their analysis with polarization microscopy but did not further attempt to simulate the observed interference patterns in respect to the dislocation type. However they confirmed interference pattern around both, pure (closed-core) screw dislocations and pipes.[43]

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The contradictions of the literature could be due to the existence of pipes with different dislocation character. In view of the considerable morphological variety of macro- and micropipes this thought must be critically considered. Furthermore, Frank's equation is only an approximation for pipes as it assumes an isotropic material as far as the elastic properties are concerned. This precondition would be met for a screw-dislocation whose line direction is perfectly parallel to c as the only stress component in this case is shear strain along $[0001]$. For deviations from this ideal orientation, the magnitude of elastic anisotropy becomes important. Anisotropy was found to be larger at higher temperatures for 3C-SiC (Ref. [44]) and 6H-SiC (Ref. [45, 46]), thus significant anisotropy must be considered during PVT growth. From the data of Kamitani et al. [35] and Eq. 5 a value of $A_{\max} \approx 16\%$ can be approximated at room temperature for the stress field of a dislocation with line and Burgers vector both lying perpendicular to the c -axis.

$$A_{\max} = 1 - \frac{c_{44}^*}{\frac{1}{2} \cdot (c_{11}^* - c_{12}^*)} \quad \text{Eq. 5}$$

In Eq. 5, c_{nm}^* is used in contrast to the general notation c_{ijkl} .

For any angle α between line direction and c -axis, Eq. 6 can be used to calculate the value for the effective anisotropy:

$$A_{\text{eff}} = \sin(\alpha) \cdot A_{\max} = \sin(\alpha) \cdot \left[1 - \frac{c_{44}^*}{\frac{1}{2} \cdot (c_{11}^* - c_{12}^*)} \right] \quad \text{Eq. 6}$$

Edge dislocations can be evaluated using the very same procedure described earlier, although the resulting equations are more elaborate as both shear and principal stress and strain must be considered.

It is questionable whether screw-dislocations or screw components are present in dislocations, which are not parallel c . The lack of pipes perpendicular to c may well be seen as

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phenomenological evidence against this. Also, it is remarkable that micro- and macropipes do only occur in materials with sheet-like structure, i.e. highly structurally anisotropic materials (SiC, ZnS, CdI₂).^[47-49] Considering that dislocations are no equilibrium phenomenon the system favours their removal and therefore they will move out of the crystal along glide planes as soon as critical activation energy for this movement can be overcome. For edge dislocations, gliding is possible perpendicular to the line direction and parallel to the Burgers vector. Gliding of screw dislocations happens perpendicular $\vec{b}$. Fig. 13 depicts a schematic illustration of possible gliding planes for 6H-SiC.

Movement of a dislocation with an integer Burgers vector requires gliding of the parts of the crystal structure by one unit cell parameter. A 6H-SiC gliding along [0001] therefore requires movement of 1.5079 nm in c- and 0.3073 nm in a-direction, i.e. moving within the a-b plane requires a “step” that is only one fifth of the step width along c. It is trivial to assume the activation energies to follow the same correlation. Moreover, in the a-b plane there exists one more path for gliding, because in $\langle 1\bar{1}00 \rangle$ it is possible for the dislocation to move in two steps along $\langle \frac{1}{2}\bar{1}\frac{1}{2}0 \rangle$ and $\langle 1\frac{1}{2}\bar{1}0 \rangle$ (= 0.2661 nm each). This way, stacking faults will temporarily be created. This mechanism may be energetically favourable, because stacking faults contain only low energies. Along c no such “shortcut” can occur as no partial shift will result in a close-packed structure.

Therefore dislocations with a Burgers vector in the basal plane are relatively mobile within the crystal structure, while in case of $\vec{b}$ parallel c dislocations are kinetically disfavoured for gliding. This means that screw-dislocations will be found parallel c and only edge dislocations can be found within the basal plane. Assuming a sufficiently high annealing temperature, pipes, which are not perfectly arranged along the c-axis must show some edge component as any pure screw dislocation component with $\vec{b}$ lying in the basal plane would have vanished.

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Gliding of dislocations with a non-integer absolute value of $\bar{b}$ will result in a high level of stacking faults. The observation of Takahashi et al. [11] of pipe-free but stacking-fault

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containing SiC crystals grown along $[\bar{1}100]$ favours the assumption that dislocations parallel to the growth direction will be created during the crystal growth on the seed crystal and anneal later, depending on the actual orientation.

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Unfortunately most published studies do not describe the exact orientation of the observed pipes relative to the c-axis. Standard wafers are cut to a thickness of 250 μm making a large-scale examination of the exact defect orientation impossible. Fig. 3 showed a typical case of a medium-grade SiC single-crystal ingot cut perpendicular to (0001). Here, the pipes propagate along the curved outer face of the crystal rather than purely along c. Some scatter of the literature value for the true character of pipes may be explained this way.

One more complication of Frank's equation is that no additional strain is considered. Clusters of dislocations within the basal plane were reported to be radially distributed around pipes [50] and to dissociate into Shockley partial dislocations. Their formation is due to the system's impetus to reduce the stress around pipes, which will influence the equation radius of the formed open-core dislocation. These considerations are even more complicated by the observation of unevenly shaped pipes, which raises even more questions on the applicability of Frank's equation.

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Considering the statements made above, the observation of a strain field of a dislocation with both screw and edge components can be explained in two ways. (1) Edge dislocations around pipes may superimpose the stress field created by the screw dislocation to yield an interference pattern similar to the one of a mixed-type dislocation stress field. This however would require a non-uniformly distributed stress field, because a homogenous stress field would cause both constructive and destructive superimposition of the stress fields (cf. Fig. 9c). (2) The studied pipes probably have both edge and screw components as they do not lie

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perfectly parallel to c. This would well explain all observations as long as the simulated data of closed-core dislocations with and without an edge component are truly extendable to open-core defects.

6. Conclusion

We reported on the variability of the general outline and interference pattern of hollow macro-defects commonly known from SiC single-crystals. Both defect types, voids and pipes, exhibit a wide range of structural variability, which accounts for the discrepancies in the literature concerning the nature of hollow-core dislocations. Those were described earlier to be pure screw-dislocations (e.g. of Ma et al. [41]), but the work Ohsato et al. [37, 38] raised serious doubts upon this interpretation, because the interference pattern were unexplainable unless a edge-component to pipes was introduced.

We confirm these findings and attribute the observed interference phenomenon to the character of pipes with an inclination towards the c-axis which always have an edge component. Furthermore, voids were shown to be in a relaxed state due to their origin from a local vapour transport growth process, even within a superimposing strain-field.

Conoscopy allows gaining more information about a general inclination of the single-crystal and especially about anomalous biaxial behaviour with undulatory extinction. Considering how quick and easy-to-apply polarisation and conoscopical microscopy is compared to other defect-characterization methods, we think that routine application of these methods e.g. for quality management will become important. Kubota et al. recently described a method to automate this way of rapid defect mapping [51]. Also, conosopic imaging allows accessing information about the angle, in which a certain wafer has been cut.

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Fig. 1: Schematic illustration of a void formation mechanism (after Ref. [12]) which also induces a micropipe to form. Note the trench at the void's outer region.
38x52mm (600 x 600 DPI)

Figure 2. Various forms of macropipes found on (0001)-SiC surfaces (SE-mode SEM microphotographs). The pipe shown in Fig. 2h is characteristic for high-quality wafers.
242x325mm (96 x 96 DPI)

Figure 3. Vertically sliced ingot showing pipes that deviate in their propagation direction significantly from the c-axis.
156x54mm (600 x 600 DPI)

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Fig. 4: Various forms of polygonized voids as observed in 6H-SiC (light microscopically images).

388x391mm (96 x 96 DPI)

Figure 5. Polarization microphotographs (a, b) and SE-mode SEM picture of voids. a) Vertically sliced void showing its preferential extension in the basal plane. b, c) Polygonized voids terminating large open-core dislocations.
315x534mm (96 x 96 DPI)

Fig. 6: Bright interference pattern around a large, elongated pipe similar to the one shown in Fig. 2b with a maximum diameter of approximately 5 μm (deviation from c-axis confirmed by focussing different levels).

368x274mm (96 x 96 DPI)

Figure 7. Section of SiC wafer containing macropipes (MP). (a) Without crossed poles and (b) crossed poles. The macropipe (MP) in the middle is easily seen by the interference pattern, while in reflected light mode those small open-core dislocations are very difficult to see. A small void (V) is only visible, because a micropipe originates from it.

471x294mm (96 x 96 DPI)

Figure 8. Schematic conoscopic images of a 6H-SiC:N-Wafer cut perpendicular c (columns a, b) and with a 3.5° inclination towards the c-axis (column c). The rotation is anticlockwise. The colours denoted in the first figure are the birefringence colours with a red I compensator (λ -plate: 551 nm) inserted. Column a): Ideal positive optical uniaxial character of 6H-SiC: centred, closed isogyres. Column b): In strained areas around e.g. micropipes, the optical uniaxial crystal becomes locally anomalous biaxial and the isogyres opens slightly. Column c): A wafer cut with an offset towards the c-axis with closed, but not centred isogyres.

67x71mm (600 x 600 DPI)

Figure 9. a) Shape of the interference pattern around open-core dislocations in crossed poles condition. b) Interference pattern with red and blue areas after introduction of a red-I-plate. c) Destructive and constructive interfering areas of the strain field around pipes in the presence of a superimposing strain field.
96x123mm (300 x 300 DPI)

a) Interference pattern of a pure screw-dislocation

b) Interference pattern of a mixed-type dislocation

Figure 10. Simulated contour shape (= lines of constant intensity) for a) a pure screw-dislocation and b) for a mixed-type dislocation (i.e. having both edge and screw components).

67x45mm (600 x 600 DPI)

Fig. 11: Comparison between observed and simulated interference pattern. The pipe studied showed only a small deviation from the c-axis as confirmed by focussing different levels within the transparent crystal.

Fig. 12: Voids as seen using crossed poles. Pipes are surrounded by significant interference patterns (a); such halos are not present around voids (b). Any brightening around voids is related to clustered screw-dislocations (c).

285x532mm (96 x 96 DPI)

Fig. 13: Schematic illustration of the crystal structure of 6H-SiC with two possible glide planes.
66x55mm (600 x 600 DPI)