A holistic X-ray analytical approach to support sensor design and fabrication: Strain and cracking analysis for wafer bonding processes

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HIGHLIGHTS

- Silicon-to-sapphire wafer bonding through an innovative method and characterized by high-resolution XRD and X-ray micro-CT.
- The innovative silicon-to-sapphire wafer bonding proved effective and strong opening the floor to a remarkable advancing in packaging design.
- X-ray micro computed tomography revealed powerful for quantifying 3D-volume defects in mm-thick structures and assess the bonding process effectiveness.
- A comprehensive understanding of the degradation mechanisms acting at the micron/nano-scale is only accessible by a conjoint X-ray-based approach.
- The approach demonstrated applies to the quantitative analysis of the status of materials involved into each partially destructive microfabrication process.

GRAPHICAL ABSTRACT

Analytical Support for MEMS Design and Fabrication

ABSTRACT

Devices such as sensors, actuators, or micro-electromechanical systems (MEMS) are obtained by a variety of microfabrication processes. Many of these processes influence the material systems by the introduction of strain and defects, which may affect the final device’s performance and reliability. Indeed, controlling materials’ status during the microfabrication is fundamental for the process optimization itself and for guaranteeing the highest devices performances during their lifetime. In this work, a conjoint analytical approach between high-resolution X-ray diffraction (HRXRD) and X-ray micro-computed tomography (CT) evaluates an innovative silicon-to-sapphire wafer bonding process. Large cracks 30–60 μm-thick were identified in both crystals by micro-CT and related to the interfacial high-stress release. In parallel, a multi-domain microstructure associated with strain and tilt affect the silicon crystallinity due to

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degradation
Stress in crystals

1. Introduction

X-ray analytical methods are known for their extraordinary strength: there are generally non-destructive for many inorganic materials and make invisible structural features, as 3D volumes, visible. X-ray diffraction (XRD) is one of the most powerful analytical techniques for the investigation of a wide range of microstructural features from bulk high-quality single crystals presenting an extremely high degree of structural ordering to quasi-amorphous nanocrystalline materials. Material properties that can precisely be evaluated by XRD are order–disorder transitions [1], chemical composition and crystalline phases, preferred orientation and texture [2], size of crystallites [3–5] micro-stress, and defect densities and mobility [5–7].

The extensive processing and utilization of bulk single crystals and high-quality epitaxial thin films by the microelectronic industry led to the improvement of the analytical performances of XRD instruments, opening the door to high-resolution X-ray diffraction (HRXRD) [7–11]. The reasoning behind this is the determination of microstructural material features that are needed to understand the microfabrication processes, i.e., deposition methods and related parameters, wafer bonding, etching, grinding, ion implantation [12]. In this concern, HRXRD was successfully exploited for the analysis of epitaxial thin films of ceramic materials for photonic applications [5,13] showing a clear impact on the optimization of the deposition parameters. Dolabella et al. have applied this method to the analysis of a new type of nanowire system to study the influence of critical steps of the fabrication process, such as electron-beam lithography and deep reactive ion etching [14]. In Yildirim et al., the microstructural investigation of nanometer-thick Pt layer revealed fundamental for the evaluation of the performances of a novel deposition method [15]. Strain and stress affecting materials lattice and their relationship with devices reliability were evaluated in Si-based MEMS for sensing applications [16] and the correlation with the wafer bonding process was assessed [17]. HRXRD is used also for the characterization of high-quality epitaxial III-nitrides films of large interest in optoelectronics, high-power and high-frequency electronics [18].

X-ray micro CT has seen a period of rapid growth over the last 20 years and is nowadays routinely applied as a commonly available tool in materials science laboratories. This is due to considerable improvements achieved in spatial resolution, in image-reconstruction time, and to the increasing availability of commercial instruments. Micro CT can deliver the 3D (volume) distribution of different material features including inclusions and matrix morphology [19], cellular morphology, porosity, defects, fibrous structure morphology [20] at single-digit micrometer scales. For recent reviews for material science applications see e.g. [21–23].

In this work, an analytical approach conjoining HRXRD and micro-CT has been exploited for the characterization of the impact of a novel wafer-bonding process in the progress of development on the material’s structure, from the nanoscale up to the micronscale. The innovative packaging itself, not being in the focus of this work, is however extremely promising in virtue of its applicability to a wide range of materials classes (ceramics, oxides, metals, nitrides) and the lowering of the process temperature below 200 °C. These advantages nominate the novel packaging approach to accompany the well-established anodic bonding technology, the latter being limited by its suitability for only silicon-to-glass bonding and much higher process temperatures.

Even if the integration of silicon-on-sapphire in CMOS (Complementary Metal Oxide Semiconductors) is promising for the development of optoelectronic devices and MEMS packaging design [24], it remains challenging. Here, single-crystal silicon and sapphire wafers are bonded together, and even if the material’s microstructure suffers so far from process development uncertainties, the effectiveness of the bonding itself is demonstrated.

At the heart of this work is the demonstration of the analytical approach: it allows to relate all the observed fabrication-related defects to the high residual stress at the bonding interface, and the consequent materials cracking by stress/strain relaxation. This appears relevant in the frame of controlling the microfabrication processes and preventing the corrosion of materials which translates into premature devices’ failure and performances impoverishment.

2. Samples preparation

Samples are fabricated employing an innovative packaging technique in the progress of development. This allows overcoming typical limitations related to the anodic bonding, such as a relatively high process temperature and the applicability to only silicon-to-glass bonding. Samples studied in this work are constituted by a Si(111) p-type 500 μm-thick wafers bonded with a c-sapphire 1000 μm-thick wafers. The bonding process is performed at low temperature (typically below 200 °C) and high voltage (typically more than 1 kV). After the bonding process, the wafer was diced to obtain squared pieces, 1x1 cm². Fig. 2 (a) shows an optical microscopy image of the sample cross-section.

3. Experimental approach

High-resolution X-ray diffraction (HRXRD) measurements were performed by means of a BRUKER D8 Discover DaVinci equipped with high-resolution optics of the primary and scattered beam, as shown in Fig. 2(b). The X-ray beam is generated by a Cs-sealed source and collimated by a Goebel mirror. A Bartels 4xGe (220) monochromator is mounted to limit the energy and angular acceptance of the primary beam to the sample. A 3xGe(220) crystal analyzer is mounted in front of the scintillator detector to reduce the angular divergence of the diffracted beam. This configuration allows an angular resolution of 0.0015°, which translates into a strain resolution of 6°10⁻⁵ on the Si(111) symmetrical reflection. In the present study, the material’s microstructure was investigated by performing scans along the radial direction, i.e. (0–2θ) scan, the angular direction, i.e. (Ω–rocking curves (RCs) and Reciprocal Space Maps (RSMs), i.e. (Ω + 0B)–2θ scans at different 0B offset. These analyses were performed on the symmetrical Si(111) reflection. The width of the line beam in the direction of the scattering plane was changed by means of the use of slits of different aper-

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tecture, 600 μm, 200 μm, and 50 μm. The width of the beam on the sample surface, L, is given by the relationship $L = \frac{w}{\sin \alpha}$, being w the nominal slit aperture and $\alpha$ the angle incident to the surface. Accordingly, L is calculated to be 2442, 814, and 204 μm, respectively. Given the extremely small divergence of the beam and high movement precision ($\pm 0.0001^\circ$), the only source of noticeable errors on these values is due to the uncertainty on the width of the slits. These high-precision optics normally have a tolerance of $\pm 1 \mu m$ over the nominal aperture, which impacts the final value of L for not more than $\pm 4 \mu m$. Being the length of the beam in the direction perpendicular to the scattering plane 1 cm, the footprint of the beam on the sample surface is calculated to be 24.42, 8.14, and 2.04 mm$^2$, respectively, when the three slits are used and the Si (111) symmetrical reflection is measured. The X-ray penetration depth is calculated according to the mass attenuation coefficient for silicon [25]. At an energy of 8 keV, X-rays travel around 70 μm into silicon [21,22], which corresponds to a penetration depth of only 17.2 μm when the symmetrical Si(111) reflection is considered. Accordingly, the irradiated volume is 0.04, 0.14, and 0.42 mm$^3$, respectively, when the three apertures are changed. The relationships between the optics implemented at the primary beam and the amounts of the sample irradiated. Accordingly, the irradiated volume is 0.04, 0.14, and 0.42 mm$^3$, respectively, when the three apertures are changed. The relationships between the optics implemented at the primary beam and the amounts of the sample irradiated.

Relationship between the optics implemented at the primary beam and the amounts of the sample irradiated.

| Width (μm) | $\alpha$ (°) | Penetration depth (μm) | L (μm) | Irradiated surface (mm$^2$) | Irradiated volume (mm$^3$) |
|-----------|--------------|------------------------|--------|---------------------------|---------------------------|
| 50        | 14.2211      | 17.2                   | 204    | 2.04                      | 0.04                      |
| 200       | 814          | 8.14                   | 814    | 0.14                      | 0.14                      |
| 600       | 2442         | 24.42                  | 2442   | 0.42                      | 0.42                      |

4. Results and discussion

In a bonding process, the participating materials undergo changes guided by the generation of mechanical stress at the bonding interface and defect state which influences final devices’ performance and reliability. A holistic interpretation of the effect of bonding by a combination of non-destructive X-ray-based methods is given. While X-ray micro CT qualifies and quantifies micron-sized (or larger) 3D (volume) defects such as voids and cracks, HRXRD quantifies strain and defects concerning the atomic structure of the respective materials. The results give us the possibility to correlate materials strain (HRXRD) and at the same time strain release through the appearance of voids and cracks. While HRXRD gives information on a 2½D level, it can be combined with full 3D information by X-ray micro CT.

4.1. X-ray micro CT: Voids and cracks analysis at the bonding interface

Three major cracks have been identified based on the micro CT images in the sample at the given spatial resolution and contrast. This is shown in Fig. 4 (a). They all lie across the bonding surface, although the red and green are predominantly in the sapphire part, whereas the blue one is almost entirely in the Si wafer.

The thickness distribution of the cracks has been determined. This is shown as a color-map distribution in the 3D space in Fig. 4(c). Note that although subpixel interpolation is applied in VG Studio Max 3.3 [26] for solid surface and wall-crack-thickness determination, thickness below the spatial resolution limit, 16 μm, is not reliably determined.

Interestingly, the elongated red crack, which mostly propagates into the sapphire crystal, trespasses into the silicon single crystal. The absence of a discontinuity between the two materials provides information about the fact that the bonding process was successful and is effective. In the case of discontinuity between silicon and sapphire, the crack would not have traveled across the interface.

4.2. HRXRD: Interface strain and defects analysis for the materials at the interface

At first, the relationship between the crystalline orientation of Si and its microstructure is evaluated. Initially, we probed a larger part of the Si crystal by irradiating around 24.42 mm$^2$ of the sample surface. Fig. 4 shows the Si microstructure being strongly dependent on the azimuthal orientation; it is, on the contrary, unaffected by the zone of samples where the analyses are performed, center or close to edges respectively. The different broadening profile is surprising, because of the theoretical in-plane isotropy of Si when the [111] is oriented parallel to the surface normal. The Si(111) Bragg peaks are broadened, mainly in the angular direction (omega relative), when the reciprocal space is collected along with the Phi 0° direction, as Fig. 5 (a) and (c) show. Fig. 5(b) and (d) show the broadening of the same lattice point when the reciprocal space is scanned along with the orthogonal azimuthal direction, namely Phi 90° (look at Fig. 1 and Fig. 3 to visualize the meaning the Phi 0° and 90°). In this case, the microstructure appears profoundly changed, with a net separation of the lattice point intensities in three crystalline domains, each characterized by a different status of strain and tilt, the latter being directly readable on the omega relative axis.
Fig. 1. (a) HRXRD experimental configuration; (b) X-ray penetration depth d in Si as a function of the incidence angle α. For the Si(111) reflection, α = 14.2211° and d = 17.2 μm.

Fig. 2. (a) Optical microscopy picture of the cross-section of the analyzed sample; (b) D8 Discover DaVinci equipped with high-resolution optics of the primary beam (Goebel mirror and Bartels monochromator) and diffracted beam (3xGe(220) crystal analyzer); (c) external of the Rx Solution Easytom XL micro/nano CT; (d) inside the cabinet of an Rx Solution Easytom XL micro/nano CT on left with the flat panel detector, on the right the nano and micro-focus X-ray tubes and in the middle below the rotating sample stage.

Fig. 6 displays the details of the RSM in Fig. 5d. Both the radial ω-2θ (or θ-2θ) scan and the angular ω-RC extrapolated along with the RSM profile point out the presence of many crystalline domains, each characterized by a different radial strain (Δd_{111}) and tilt to the surface normal. The extrapolated 1D profiles were fitted using three Gaussian profiles. Even if it represents an approximation of the real status of the crystal, the position of these peaks on the 2 theta and omega relative scales can provide information about the gradients of strain and tilt, respectively.

The strain normal to the parallel Bragg plane is obtained by X-ray diffraction methods from the deviation of the Bragg angle from the reference position, according to the following calculation:
Fig. 3. Micro-CT 3D-rendering of silicon (yellow) and sapphire (grey). The azimuthal directions along with HRXRD measurements were performed, namely Phi 0° and Phi 90° respectively, are explicated. The zones of silicon probed by the X-ray beam, namely “Probed volume 1” and “Probed volume 2” respectively, are indicated for both the azimuthal orientation, (a) Phi 90° and (b) Phi 0°. (For interpretation of the references to colour in this figure legend, the reader is referred to the web version of this article.)

Fig. 4. (a) 3D rendering of the micro-CT image of the full Si-sapphire sample shows the 3 cracks found in the sample (segmented in red, green blue). The Si wafer is shown in light yellow and the sapphire is in semi-transparent light grey. (b) shows the parts of the cracks penetrating the Si wafer by setting the sapphire non-transparent. (c) thickness distribution of the blue crack. The mean crack thickness is 42 μm. (For interpretation of the references to colour in this figure legend, the reader is referred to the web version of this article.)
being \( d \) and \( d_0 \) equal to \( d = \frac{n}{k_2 \sin \theta} \) and \( d_0 = \frac{n}{k_2 \sin \theta_0} \), respectively, the spacing for a certain family of atomic planes \( \text{(hkl)} \), and \( \theta \) and \( \theta_0 \) the measured and reference Bragg angles, respectively.

Values of strain and tilt and the lattice parameter are reported in Table 2. The tilt of each crystalline domain was double-checked by plotting the scattering vector components, i.e. \( q_x \) and \( q_z \), in the reciprocal space according to the equation:

\[
\text{tilt}(^\circ) = -\arctg \left( \frac{q_x}{q_z} \right)
\]

The peak 1 is located at the position \( 0^\circ \) in the omega relative scale and is considered as a reference. This is assumed as the alignment was performed on this peak. The tilting of the other crystalline domains is consequently reported with respect to the reference peak.

As visible in Fig. 5, the silicon microstructure does not change when moving from the edge (a and b) to the center (c and d) of samples. The same values of tilt and strain were extracted from both probed points. It means that the Si microstructure is homogeneous over the entire sample and it is not influenced by the vicinity of the diced edge. Even if this information is not central for this work, it presents a piece of interesting information provided by our analytical approach. Accordingly, the defects are not provoked by the dicing of the bonded wafers, but by the bonding process itself. The fact that the microstructure of silicon does not change if considered at the edges or the center of the sample, likely implies that the microstructure degradation is not a direct consequence of the large red crack shown in Figs. 3 and 4, which is located on aside. More likely, the degradation affecting the Si microstructure has to be related to the stress generated at the bonding interface which travels through the crystal. The multi-domain microstructure is then attributed to the formation of small cracks to relieve the high bonding stress. Indeed, the part of the crystal close to these cracks relaxes the lattice strain. On the contrary, the parts of the crystal farther from the cracks present residual stress, which translates into lattice strain. These cracks have most probably dimensions down to the resolution limit of the imaging technique.
Tilt and strain of the crystalline domains visible in the Si(111) RSM in Fig. 4.

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The penetration depth of 8 keV X-rays into Si is less than 14.22 μm, meaning that cracks and defects generated at the bonding interface due to the high stress can propagate along the entire crystal until the sapphire crystal. There is no discontinuity at the bonding interface, as shown in Fig. 7(c); but also, in this case, these cracks likely originate from the crystals and propagate from the crystal inner part to the bonding interface. However, the fact that cracks propagate from one crystal (sapphire) to the other (silicon), observed in both micro-CT (red crack in Fig. 3) and SEM (Fig. 6(c)), points out the strength and effectiveness of the bonding. This observation is further underpinned by observing the dicing-related cracks (appearing as lines in the SEM images) propagating in the direction normal to the Si surface. The generation and propagation direction of these cracks depend on the crystal orientation as the elastic properties in crystals are direction-dependent and the direction of dicing stress application. These defect lines propagate normal to the bonding interface in the vicinity of the Si surface and along the Si crystal. When the bonding interface is approaching, the propagation direction changes, and cracks propagate into the sapphire crystal. There is no discontinuity at the bonding interface. The deviation of these cracks is probably related to the stress field at the interface due to the bonding process.

An interesting point is related to the observation of the degradation of the microstructure in the most outer part of the silicon crystal. The penetration depth of 8 keV X-rays into Si is less than 20 μm when the 111 reflection is considered, the incidence angle being only 14.22°, as it is possible to extrapolate from Fig. 1(b). It means that cracks and defects generated at the bonding interface due to the high stress can propagate along the entire crystal until the surface. The sample cross-section was also investigated by scanning electron microscopy (SEM) to visualize the status of the bonding interface and the presence of visible cracks. Fig. 6(a) and 6(b) show that most of the bonding interface does not show noticeable cracks. At least one large crack affects the Si crystal, but its relationship to the bonding process or the cutting of the wafers after bonding cannot be established. Only a small part of the bonding interface presents cracking, as shown in Fig. 7(c); but also, in this case, these cracks likely originate from the crystals and propagate from the crystal inner part to the bonding interface. However, the fact that cracks propagate from one crystal (sapphire) to the other (silicon), observed in both micro-CT (red crack in Fig. 3) and SEM (Fig. 6(c)), points out the strength and effectiveness of the bonding. This observation is further underpinned by observing the dicing-related cracks (appearing as lines in the SEM images) propagating in the direction normal to the Si surface. The generation and propagation direction of these cracks depend on the crystal orientation as the elastic properties in crystals are direction-dependent and the direction of dicing stress application. These defect lines propagate normal to the bonding interface in the vicinity of the Si surface and along the Si crystal. When the bonding interface is approaching, the propagation direction changes, and cracks propagate into the sapphire crystal. There is no discontinuity at the bonding interface. The deviation of these cracks is probably related to the stress field at the interface due to the bonding process.

To understand the origin of this microstructure degradation observed in the direction Phi 90°, we also varied the volume of crystal probed by the X-ray beam. The size of the footprint in the horizontal direction is particularly relevant because this direction coincides with the scattering plane. The effect of the divergence in this direction is supposed to be negligible as it is limited to only 7 arcsec by the use of the Bartels monochromator in the incident beam path. Consequently, the variation of the size of the line-focused beam footprint in the horizontal direction only translates into a variation of the probed volume of the crystal.

The effect of the variation of irradiated volume of Si on the tilting dispersion (shape and width of the RCs) is dramatic. Together with the expected decrease of the intensities related to the less exposed sample surface, one can observe a change in the widths and shapes of the rocking curves. When the irradiated surface is 24.42 mm² (blue curve in Fig. 7(d)), the RC analysis reveals a relatively large spread of the Si(111) atomic plane tilt distribution, over 0.10°. The curve appears as a broad distribution of many tilted crystal components instead of the sharp peak profile expected from a high-quality Si single crystal diffraction peak. The sample shows many crystalline domains tilted to each other. When reducing the irradiated surface to 8.14 mm² (red curve in Fig. 7(d)), the tilt distribution profile is less broad, with mainly two crystalline components appearing. This is further reduced if the irradiated surface is 2.04 mm² (black curve in Fig. 7(d)). Here, the broad profiles are substituted by three peaks, pointing out three crystalline domains tilted to each other and the surface normal. This HRXRD analysis shows the evidence of the fragmented microstructure of Si after the wafer bonding process: the larger the probed sample surface the higher the broader the distribution of tilted domains that are in diffraction conditions.

The analysis of the RCs changes reveals that the size of single-crystalline domains in the investigated volumes is most probably very similar. When irradiating a smaller part of the sample, we investigate only a few domains showing smaller tilts of about 0.03° to each other; we can capture individual Si(111) domains with an approximate FWHM of about 0.02°. Going to larger probed sample volumes, we observe a superposition of the reflections of individual tilted crystalline domains which results in a very broad rocking curve.

The evidence of the fragmented microstructure appears clearly visible in the profile of the Si(111) RSM in Fig. 6(a). The profile of the Si(111) RSM in Fig. 6(a) shows the evidence of the fragmented microstructure of Si after the wafer bonding process: the larger the probed sample surface the higher the broader the distribution of tilted domains that are in diffraction conditions.

An interesting point is related to the observation of the degradation of the microstructure in the most outer part of the silicon crystal. The penetration depth of 8 keV X-rays into Si is less than 20 μm when the 111 reflection is considered, the incidence angle being only 14.22°, as it is possible to extrapolate from Fig. 1(b). It means that cracks and defects generated at the bonding interface due to the high stress can propagate along the entire crystal until the surface. The sample cross-section was also investigated by scanning electron microscopy (SEM) to visualize the status of the bonding interface and the presence of visible cracks. Fig. 6(a) and 6(b) show that most of the bonding interface does not show noticeable cracks. At least one large crack affects the Si crystal, but its relationship to the bonding process or the cutting of the wafers after bonding cannot be established. Only a small part of the bonding interface presents cracking, as shown in Fig. 7(c); but also, in this case, these cracks likely originate from the crystals and propagate from the crystal inner part to the bonding interface. However, the fact that cracks propagate from one crystal (sapphire) to the other (silicon), observed in both micro-CT (red crack in Fig. 3) and SEM (Fig. 6(c)), points out the strength and effectiveness of the bonding. This observation is further underpinned by observing the dicing-related cracks (appearing as lines in the SEM images) propagating in the direction normal to the Si surface. The generation and propagation direction of these cracks depend on the crystal orientation as the elastic properties in crystals are direction-dependent and the direction of dicing stress application. These defect lines propagate normal to the bonding interface in the vicinity of the Si surface and along the Si crystal. When the bonding interface is approaching, the propagation direction changes, and cracks propagate into the sapphire crystal. There is no discontinuity at the bonding interface. The deviation of these cracks is probably related to the stress field at the interface due to the bonding process.

Table 2
Tilt and strain of the crystalline domains visible in the Si(111) RSM in Fig. 4.

| Peak | Tilt (°) | Strain d_{111} (10^{-5}) | Strain d_{111} (%) |
|------|---------|--------------------------|--------------------|
| 1    | 0.002   | NA                       | NA                 |
| 2    | 0.021   | 2.6                      | 0.026              |
| 3    | 0.059   | 7.0                      | 0.070              |

Fig. 6. (a) Profile (black line) of the Si(111) RSM in Fig. 1d. (b) On the top: ω-2θ scan along with the profile; on the bottom the omega relative scan along the profile.
which migrate along the entire thickness of the silicon crystal until the surface. When the stress overcomes a certain threshold, we observe the generation of larger cracks through micro CT analysis in the vicinity of the bonding interface. This is a piece of unique information that can be provided by micro CT only, because of the inaccessibility of the near bonding interface region by the 8 KeV X-rays used in HRXRD. At the same time, together with the latter macro-cracking, we observe crystalline defects due to the breaking of the initial perfect crystal into tilted and strained domains, involving also the formation of micro-cracks with size likely down to the imaging technique resolution.

5. Conclusions

In this work, we followed a holistic analytical approach using HRXRD and X-ray micro-CT to investigate the impact of an innovative wafer bonding process on the microstructure of p-type SC-Si. The 3D images obtained by X-ray micro CT point out the presence of large cracks and voids affecting the near bonding interface region. The imaging technique also proved the effectiveness of the bonding process between silicon and SC-sapphire by the absence of discontinuity between the two bonded crystals at the interface. One of the identified cracks travels into the sapphire crystal and then propagates into the silicon. This was confirmed by SEM showing the bonding interface homogeneous mostly without cracks. This represents an outstanding result due to the promising integration of silicon-on-sapphire CMOS for the optoelectronic industry. The HRXRD characterization allowed the determination of tilt gradients of the Si 111 atomic plane and the micro-strain along the radial [111] crystalline direction. The gradient of the strain and the presence of many crystalline domains characterized by a different tilt and strain can be correlated with the generation of cracks through bonding stress relaxation. These cracks have dimensions below the resolution limit of the imaging technique and travel along silicon until reaching the most outer part of the crystal. The diffraction technique is not susceptible to reveal the effects of the large cracks at the bonding interface, which are on the contrary, clearly detected and quantitatively characterized by the imaging technique. Additionally, at the energy employed by conventional Cu-radiation X-ray diffractometers, the thickness of Si probed is limited to a few tens of microns. X-ray micro CT allows, on the contrary, the observation of the whole thickness.

The obtained results show the strengths in combining HRXRD, SEM, and micro-CT in a holistic approach to provide complementary strain and defects information about the sample. In this frame,
our analytical concept represents a powerful investigation able to reveal different types of structural degradation mechanisms, which are not accessible with any other analytical approach. The success of the obtained bonding points out the effectiveness of the novel packaging approach here reported for the first time; related to the versatility, its optimization may represent outstanding progress in the field of devices' design.

Declaration of Competing Interest

The authors declare that they have no known competing financial interests or personal relationships that could have appeared to influence the work reported in this paper.

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