On the Location of Boron in SiO$_2$-Embedded Si Nanocrystals—An X-ray Absorption Spectroscopy and Density Functional Theory Study

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1. Introduction

The incorporation of boron (B) impurities into ultrasmall silicon nanocrystals (Si NCs) and other Si nanovolumes is technically not trivial and its efficient feasibility is still a matter of debate. The first reports about B-doped Si NCs attributed photoluminescence (PL) quenching to nonradiative Auger recombination of excitons with doping-induced free carriers.[1] Meanwhile, several reports support the concept that Si NCs ≤6 nm in diameter are not electronically p-type-doped (i.e., possess free carriers from B dopants) but that B in the NCs or the surrounding SiO$_2$ creates nonradiative defect states.[2–5] The situation is further complicated by the preferential incorporation of B on or near the surface of Si NCs as demonstrated in theoretical and experimental work.[3,6–11] In such configurations, the substitutional incorporation of B on Si lattice sites with exclusive first next-neighbor (1nn) Si atoms as a mandatory requirement for electronic doping (e.g., not observed for phosphorus) that adds to the established nanosize effects, namely, increased dopant activation and ionization energies.

Doping of silicon nanostructures is crucial to understand their properties and to enhance their potential in various fields of application. Herein, SiO$_2$-embedded Si nanocrystals (quantum dots) ≈3–6 nm in diameter are used as a model system to study the incorporation of B dopants by X-ray absorption near-edge spectroscopy (XANES). Such samples represent a model system for ultimately scaled, 3D-confined Si nanovolumes. The analysis is complemented by real-space density functional theory to calculate the 1s (K shell) electron binding energies of B in 11 different, thermodynamically stable configurations of the Si/SiO$_2$/SiO$_2$ system. Although no indications for a substitutional B-acceptor configuration are found, the predominant O coordination of B indicates the preferred B incorporation into the SiO$_2$ matrix and near the Si-nanocrystal/SiO$_2$ interface, which is inherently incompatible with charge carrier generation by dopants. It is concluded that B doping of ultrasmall Si nanostructures fails due to a lack of B incorporation onto Si lattice sites that cannot be overcome by increasing the B concentration. The inability to efficiently insert B into Si nanovolumes appears to be a boron-specific fundamental obstacle for electronic doping (e.g., not observed for phosphorus) that adds to the established nanosize effects, namely, increased dopant activation and ionization energies.

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concentrations in Si NCs (hyperdoping).\textsuperscript{[12–14]} However, such B concentrations induce a phase transition from B-doped Si to Si borides.\textsuperscript{[15]}

In this work, we study in theory and experiment the chemical environment of B in the Si NC/SiO\textsubscript{2} system to investigate the local atomic surroundings of B atoms and the preferential sites for their incorporation. For this purpose, we combine X-ray absorption near-edge spectroscopy (XANES) at the B K-edge with density functional theory (DFT).

2. Results

Table 1 gives an overview on the configuration of all B-doped Si NC/SiO\textsubscript{2} samples investigated here as well as the B-doped SiO\textsubscript{2} and bulk-Si reference samples.

Figure 1 shows PL spectra of the Si NC samples. For the nominally 3 nm NC-size samples PL peaks at 765 nm are observed, irrespective of low (0.47 at\%) or high (1.32 at\%) B concentration. In contrast, the nominally 6 nm NC-size samples with low B concentration have a typical peak position at 905 nm, which shifts to 855 nm for high B concentration. We attribute this PL blueshift to a slightly lower excess-Si concentration in highly B-doped Si-rich oxide layers (cf., secondary ion mass spectrometry (SIMS) quantification data in Hiller et al.\textsuperscript{[3]}), which causes a slightly reduced average Si NC size. It was also shown that high B concentrations in SRO precursor layers hamper the NC formation during annealing, resulting in a reduced NC size.\textsuperscript{[16]} In addition, the preferential incorporation of near-interfacial B atoms into larger NCs (as recently reported by Demoulin et al.\textsuperscript{[11]}) could also contribute to the PL blueshift when the formation of B-induced luminescence quenching defects is considered.\textsuperscript{[3]} As expected, no PL signal is measured for the SiO\textsubscript{2}:B reference samples that do not contain any Si NCs (black curves in Figure 1).

The PL intensities of both NC sizes (not shown) drop to \(\approx 14\%\) when going from low to high B concentrations. As shown before, this PL loss is mainly attributed to B-induced defects and is not an indication of Auger quenching by B-induced free carriers.\textsuperscript{[13,14]}

For XANES measurements, we initially used a highly B-doped Si wafer reference sample (0.005–0.015 \(\Omega\) cm, corresponding to a B concentration of \(1.2 \times 10^{19}–3 \times 10^{18} \text{ cm}^{-3}\)) but failed to reveal any significant X-ray absorption peaks. Therefore, a Si wafer was B implanted to obtain a higher B concentration. In Figure 2 a depth profile of the active B concentration in sample Si:B is plotted, which shows an \(\approx 200\text{ nm deep plateau under the surface with} \approx 1 \times 10^{20} \text{ cm}^{-3}\), before its slow decline. We identify this active B concentration with the total B concentration because the postimplantation anneal was conducted at 1000 \(^\circ\text{C}\) for

![Figure 1. Normalized room-temperature PL spectra of B-incorporating Si NC/SiO\textsubscript{2} samples and SiO\textsubscript{2}:B reference samples. The suffixes -L and -H denote the low (-L) and high (-H) B concentrations.](image)

![Figure 2. Electrochemical capacitance–voltage (ECV) depth profile of the active B concentration in the B-implanted Si-wafer reference sample (Si:B).](image)

| Sample | Configuration | Avg. [B] [at\%] |
|--------|---------------|-----------------|
| 3 nm-L | 50 \((\text{3 nm SiO}_2 + \text{3 nm SRO:B @ 0.47 at\%})\) | 0.24 |
| 6 nm-L | 50 \((\text{3 nm SiO}_2 + \text{6 nm SRO:B @ 0.47 at\%})\) | 0.31 |
| 3 nm-H | 50 \((\text{3 nm SiO}_2 + \text{3 nm SRO:B @ 1.32 at\%})\) | 0.66 |
| 6 nm-H | 50 \((\text{3 nm SiO}_2 + \text{6 nm SRO:B @ 1.32 at\%})\) | 0.88 |
| SiO\textsubscript{2}:B-L | 225 nm SiO\textsubscript{2}:B @ 0.25 at\% | 0.25 |
| SiO\textsubscript{2}:B-H | 225 nm SiO\textsubscript{2}:B @ 0.8 at\% | 0.8 |
| Si:B | B-implanted Si wafer (0.2 at\% near surface; 19 \(\Omega\) cm\(^{-1}\)) | 0.2 |
The glasses This observation is another indication
Boron K-edge X-ray absorption spectra (© 2021 The Authors. Physica Status Solidi B published by Wiley-VCH GmbH samples with high B concentration. Thick black/C25 2 nm thick) native surface oxide, (B) derived from DFT calculations and calibrated to experiment; 20118–20120 Therefore, this peak marks the onset of B clustering, which implicates that a significant fraction of boron is present in a configuration that is inherently incompatible with B doping. Concerning the peak intensities, an ≈10 times higher fluorescence yield is observed for samples 6 nm-H compared to 3 nm-H, although the B concentration in the initial SRO layers was identical. However, the number of Si atoms constituting the nominally 3 nm diameter Si NCs is just 1/8 of the nominally 6 nm Si NCs. At a constant B concentration, there is therefore an eight times lower probability for B-dimer formation, which explains the peak intensity difference. Furthermore, cluster analysis of atom probe tomography data[9] demonstrates that the number of B impurities in Si NCs scales with the NC size so that at least two B atoms in one NC (prerequisite for a dimer) are predominantly found in larger NCs.

As shown more clearly in Figure 3a, only the Si NC samples with high B concentration (6 nm-H, 3 nm-H to some degree) exhibit shoulders at 193.4 and 193.65 eV, which we attribute to B at the Si NC/SiO2 interface or the Si-rich oxide transition shell around the Si NCs; see below for the assignment of XANES peaks to core level energies from DFT and associated discussion.

The most intense peak found in all samples is located at 194.25 eV. Even the sample set with low B concentration (3 nm-L, 6 nm-L, SiO2:B-L) clearly shows this peak. However, in the Si:B reference sample that peak only reaches a fluorescence yield that is at least one order of magnitude lower than for the other samples, despite comparable B concentration (cf., Table 1). Also, that peak occurs for the SiO2:B reference samples with comparable intensity as for the B-doped Si NC/SiO2 samples, which strongly suggests that it corresponds to B in SiO2. Thus, in the Si:B reference sample the peak originates from B atoms in the (typically ≈2 nm thick) native surface oxide, which explains its low intensity as compared to the >200 nm thick deposited layers. The attribution of the 194.25 eV peak to tetrahedral O-coordinated B atoms in SiO2 is supported by X-ray absorption spectra of B2O3−SiO2 glasses[20] or B-zeolites B(OSi)4 and B(OSi)3.[21]

All the deposited samples with high B concentration (3 nm-H, 6 nm-H, SiO2:B-H) show two weak and broad features at 200 and 204 eV, which have been attributed to tetrahedral and trigonal B2O3 phases before.[21,22] This observation is another indication for the segregation of boron into new phases in sample set “H” with high B concentration, though its occurrence is scarce as compared to B incorporation into SiO2.

Figure 3. Boron K-edge X-ray absorption spectra (fluorescence yield). a) Spectra of all samples offset by 0.5 each for the sake of better visibility. The dashed lines indicate the peaks discussed in the text. b) Zoomed-in and unstacked region on the low-energy side of the most prominent 194.25 eV peak of all “H” samples with high B concentration. Thick black lines show peaks and shoulders of XANES signals Eγ (B), thin black lines show Eγ(1) (B) derived from DFT calculations and calibrated to experiment; see text for details. Combined number-letter labels refer to respective graphs in Figure 4, 5, and 7, and to respective rows in Table 2.

3. Discussion

DFT calculations with Gaussian-type all-electron molecular orbital basis sets (MO-BSSs) present a vital theoretical method to interpret XANES spectra.[23] Unlike in effective core-potential (ECP) DFT methods, where at least the innermost full electron
Rochow electronegativities). In contrast, the atomic radius of Si in a hybridization (tetrahedral B coordination), and to

Another reason for the complexity of XANES spectra on B configurations in Si and related oxides is given by the more diverse locations of B in SiO₂. The ionic radius of B on an O site is close to the associated radius of O itself, viz., \( r_B > r_O \) for \( r_B \approx 0.04 \) Å versus \( r_O \approx 0.12 \) Å. In contrast, the atomic radius of Si in SiO₂ is \( r_Si \approx 1.26 \) Å, whereas B on a Si site has merely \( 1/3 \) of the atomic radius with \( r_B \approx 0.45 \) Å. These ratios of atomic radii combined with the higher Allred–Rochow electronegativity (EN) of B versus Si (2.01 vs 1.74) enables B to occupy Si as well as O sites in SiO₂, which has been demonstrated by B-doped SiO₂ characterized by capacitance–voltage (CV) and deep level transient spectroscopy (DLTS) scans in compound with DFT calculations. As for P, its atomic radius as donor on a Si site in SiO₂ differs from the

| \( E_{\text{bind}} \) [eV] | \( Q \) (B) [e] | Approximate (host lattice, site)/sample, specific configurations of 1nn atoms | Figure | Corresponding peak position observed in XANES [eV] |
|---|---|---|---|---|
| 191.07 | -0.15 | B (SiO₂, O site) | 4b | 191.5 |
| 191.86 | -0.12 | B (SiO, O site) | 4d | 191.7 |
| 191.99 | -0.09 | >B–B< (Si), 2x DB–Si | 5a | 192.0 |
| 192.04 | +0.36 | >B–OH (Si), substituting Si–Si | 7b | 192.0 |
| 192.34 | -0.19 | ≡B (Si), 1x DB–Si; conventional acceptor | 5d | 192.35 |
| 192.35 | -0.04 | >B–B< (Si), 1x DB–Si, 1x H–Si | 5b | 192.35 |
| 192.37 | -0.04 | >B–B< (Si), 2x H–Si | 5c | 192.35 |
| 193.15 | +0.56 | B (Si), 2x O–Si, 1x −Si, 1x DB–Si | 7a | 193.4 |
| 193.16 | +0.35 | B (SiO₂, Si site), 2x O–Si, 1x −Si, 1x DB–Si | 4c | 193.4 |
| 193.48 | +0.36 | interstitial O=O–B (SiO₂ shell around NCs) | 7c | 193.4; 193.65 |
| 194.25 | +0.78 | B (SiO₂, Si site); confirmed in experiment²¹,²² | 4a | 194.25 |

|²¹No articulate peak; minor possible peak signal at sample 6 nm-H only.

No articulate peak; minor possible peak signal at sample 6 nm-H only.
value of neutral Si by a considerable 44%, viz., $r_p^{0.15} = 1.08 \, \text{Å} = 1.44 \, r_S^{1.216} = 0.75 \, \text{Å}$. Although an occupation of an O lattice site by P in SiO or SiO$_2$ is already unlikely due to its pentavalent bond configuration, its substantial oxidation enthalpy of $2 \, \text{P} + \frac{1}{2} \, \text{O}_2 \rightarrow \text{P}_2\text{O}_5$ of 1493 kJ mol$^{-1}$ (7.737 eV per P atom) and its high diffusion coefficient in Si$^{25}$ allow P to form its own native oxide, reducing SiO$_2$ or SiO in the process. As for XANES, such findings are corroborated by the mass signal of the native P oxide P$_2$O$_5$ dwarning the signal for P on Si sites in SiO$_2$, which strongly suggests the oxide phase separation of P mentioned earlier.$^{[25]}$ The straightforward integration of B into SiO$_2$—and into other O-related B configurations discussed later—is corroborated by depth profiles of Si NC/SiO$_2$ superlattices (SLs) obtained from atom probe tomography (APT), where the majority of B atoms were located within SiO$_2$ and at the NC interface.$^{[9,11]}$ As we shall see later, the phase separation of B in SiO$_2$ versus the formation of its native oxide B$_2$O$_3$ behaves the opposite way, with an oxide phase separation occurring only to a minor degree. Out of the 20 approximants calculated by DFT, we present the 11 relevant ones subsequently. Where instructive, we refer to the remaining 9 approximants in the context of the 11 approximants presented here. We have to conduct a calibration of 1s$_1$ binding energies of B—E$_{1s1}$ (B)—to the peaks of XANES spectra presenting the binding energy of the K shell of B (E$_K$ (B)) in various electronic environments. We use the most prominent XANES peak of B on a Si site in SiO$_2$ at E$_K$ (B, SiO$_2$) to align E$_{1s1}$ (B) in the respective approximant (see, e.g., Figure 4a), arriving at a calibration factor of 1.042 for all E$_{1s1}$ (B) from DFT to match E$_K$ (B) obtained by XANES. Such a calibration is also known from the alignment of Raman- and infrared (IR) frequencies obtained by DFT to respective experimental spectra,$^{[27,28]}$ accounting for the approximate solution delivered inherently by DFT.$^{[29]}$ All values of E$_{1s1}$ (B), the respective atomic environment and bond geometry, and the associated E$_K$ (B) are shown in Table 2. From Figure 3 it emerges that the assignment of E$_{1s1}$ (B) values obtained by DFT to E$_K$ (B) in the XANES spectra is further complicated by several E$_{1s1}$ (B) values being very close to each other. Hence, we conducted a careful discrimination of such E$_{1s1}$ (B) values against their likely or experimentally documented occurrence in specific atomic environments (samples). However, a full deconvolution of DFT-backed XANES data as successfully conducted for P in the Si/SiO/SiO$_2$ system$^{[24]}$ does not appear to be feasible for B.

We first evaluate our DFT results of B in SiO and SiO$_2$. As mentioned earlier, B can occupy Si as well as O lattice sites in both oxides, providing four atomic configurations (see Figure 4).

As discussed before, a phase separation of B$_2$O$_3$ versus SiO$_2$B should hardly occur. This statement is corroborated by the low intensity of XANES peaks of B in borosilicate glasses and B$_2$O$_3$, occurring at $E_K$ (B) = 200 and 204 eV, respectively (see Figure 3a). The formation of BO as an intermediate oxidation product$^{[26]}$ is promoted by an O deficiency due to B being added to SiO (NC samples) or SiO$_2$ (sample SiO$_2$B) without additional O to form a commensurate oxide, viz., B$_2$O$_3$. In terms of oxidation enthalpy, the value of B$_2$O$_3$ (4B + 3O$_2$ → 2B$_2$O$_3$ + 2547.2 kJ mol$^{-1}$, equivalent to 6.600 eV per B atom) does not suffice to induce a redox reaction along 2B + 6SiO$_2$ → B$_2$O$_3$ + 6SiO in analogy to P$^{[30]}$ preventing B$_2$O$_3$ formation on a major scale altogether. In addition, the low diffusion constant of B in SiO$_2$ even at high temperature$^{[9]}$ hampers the formation of B$_2$O$_3$ on a major scale as the nominal B concentration is too low for a true alloy phase to form those oxides.

From the magnified XANES spectra in Figure 3b, we see that B on an O site in SiO$_2$ (structure 4b) virtually does not occur, and is vastly outrun by the signal of B on a Si site in SiO$_2$ (structure 4a) for all samples containing SiO$_2$ in any form. We can therefore state that B on a lattice site in SiO$_2$ nearly always substitutes Si. A rather balanced ratio of B on O versus Si sites in SiO$_2$ as found by CV, DLTS, and DFT is based on an the electronically active B concentration in SiO$_2$ which is severely limited by a Coulomb blockade.$^{[25]}$ The true physical B concentration on such sites in SiO$_2$ could thus not be characterized in the mentioned work. In contrast, DLTS is extremely sensitive (detecting sheet charge densities of $\approx 10^{14}$ cm$^{-2}$), exceeding the detection limit of XANES (sheet densities of $\approx 10^{15}$ cm$^{-2}$ when taking $\approx$1% of the integral of the B density over depth in Figure 2, corresponding to $\approx$0.01 at%) by a factor of $\approx 10^4$, enabling DLTS to characterize such B configurations that cannot be detected by XANES. The B configurations in SiO will be discussed subsequently along with other O-related B configurations occurring in the suboxide shell around the pure Si NC.

We now look at B in Si, addressing both Si NCs and bulk Si. Due to the strong localization of the B 1s$_1$ electrons, their binding energy $E_{1s1}$ (B) depends mainly on the 1nn atoms of B, with more distant host atoms only having a minor influence on

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**Figure 4.** Relevant regions of DFT approximants showing B in SiO$_2$ and SiO. Atom colors are red for O, gray for Si, white for H, and pink for B. 1nn Si atoms of B are highlighted in cyan as a guide for atomic structure. a) B on Si site in SiO$_2$, featured by a Si$_2$BO$_{11}$H$_{15}$ approximant. b) B on O site in SiO$_2$, featured by a Si$_4$BO$_{11}$H$_{56}$ approximant. c) B on Si site in SiO, featured by a Si$_9$BO$_{50}$H$_{64}$ approximant. d) B on O site in SiO, featured by a Si$_{2x}$BO$_{85}$H$_{64}$ approximant.

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E_{1s,1}(B). Dimers of the form \( \mathrm{B-B} \langle \rangle \) and \( \langle \rangle \) stand for bridge bonds between Si atoms) are readily formed within Si at high B concentrations.\(^{18-20}\) We consider three \( >\mathrm{B-B}< \) configurations in Si, differing in the passivation of the two dangling bonds (DBs) of the two 1nn Si atoms not bonded to B: fully defective (2×1nn Si-DB, structure 5a), partially passivated (1nn Si-DB and 1nn Si-H, structure 5b), and fully passivated (2×1nn Si-H, structure 5c) (see Figure 5a–c, respectively). From Figure 6b, we see that the fully defective \( >\mathrm{B-B}< \) configuration has a value of \( E_{1s,1}(B) \) that is \( \approx0.37\text{ eV} \) lower when compared to the partially and fully passivated \( >\mathrm{B-B}< \) configuration. Hydrogen (H) has an EN of 2.2 and is thus only marginally more anionic to Si (EN = 1.74) than B (EN = 2.01).\(^{26}\) However, the crucial factor is the electron affinity \( X \), with \( X(B) = -0.277\text{ eV} > X(H) = -0.756\text{ eV}.\(^{26}\) Therefore, H delivers a slightly higher electron relaxation when compared to hydrocarbon, resulting in an electron flow away from the B dimer toward the H atom(s) passivating the Si DB(s). Consequently, the partially and fully passivated \( >\mathrm{B-B}< \) configurations (structures 5b and 5c) ionize B atoms slightly less negatively when compared to the unpassivated \( >\mathrm{B-B}< \) configuration (structure 5a); see atomic charges of B in Table 2. This change in ionization of the B atoms affects the \( E_{1s,1}(B) \) level via a diminished Coulomb repulsion term within the nominal cumulative electron shell \( (1s^\uparrow,2s^\uparrow,2p^\uparrow) \), corresponding to a shift of \( E_{1s,1}(B) \) by \( \approx0.37\text{ eV}. \) Unfortunately, the \( E_{1s,1}(B) \) of this B dimer configuration nearly coincides with the \( E_{1s,1}(B) \) of a \( >\mathrm{B-OH} \) bridge substituting a bond Si-Si; see structure 7b in Table 2. Therefore, it is impossible to discriminate between both B configurations when using merely respective \( E_{1s,1}(B) \) values.
passivated >B—B< configurations, therefore not allowing an assignment to either B configuration from a theoretical viewpoint. However, because the experimental verification of the conventional B acceptor configuration in bulk Si (sample Si:B) by XANES led to only a minute bump signal as mentioned in the Results section, we assume that the XANES peak we obtained is overwhelmingly due to partially/fully H-passivated B dimers. In that context, we would like to point out that B densities of ≥0.3 at% or 1.5 x 10¹⁶ cm⁻³ in bulk Si convert the electronic structure of the latter into a semimetal.\[^{31}\]

The >B—B< configurations shown in Figure 5 introduce a considerable amount of local lattice distortion, mainly by the bonds B—B (≥1.75 Å) and B—Si bonds (≥2.13 Å) being notably smaller to the nominal Si—Si bonds (≥2.38 Å). It is therefore conceivable that a reactive species such as O can participate in alternative B configurations by breaking the bond B—Si or Si—Si at high temperature as brought about by a dopant activation or Si segregation anneal of SiO to form Si NCs in SiO₂. A single B configuration in analogy to the conventional B acceptor—cf., Figure 5d—can be established with less lattice strain and considerable energy gain by the oxidation enthalpy brought about by two bonds B—Si substituted by O bridge bonds (>O), yielding ≡Si—B(—O—Si≡₂)ₓDB—Si≡ as shown in Figure 7a. At E₁₅₁₁ (B) = 193.15 eV, there is a shoulder in the XANES spectrum of sample 6 nm-H, which is ≥0.25 eV shifted to higher Eₚ (B), whereas only one broad small peak is visible for sample 3 nm-H. The XANES signal of sample 3 nm-H also appears to contain the other two B configurations discussed later. From Figure 3b, we can derive that B configurations of the type shown in Figure 7a occur only in highly B-doped NC samples, whereby the associated XANES peak gets more articulate with increasing NC diameter. A relation clearly corresponds with the thickening of the oxide (SiO₂) shell around the NC, which is at least 6 Å thick\[^{32}\] and increases with NC diameter.\[^{33,34}\] We can thus locate the ≡Si—B(—O—Si≡₂)ₓDB—Si≡ configuration to the SiO₂ shell with small x values where a sizeable amount of Si—Si bonds is still present, thus at the transition to the pure Si NC lattice. The associated XANES peak grows with NC diameter simply by the corresponding increase in NC surface. Another configuration that bridges Si—Si bonds is given by >B—OH at binding energy E₁₅₁₁ (B) = 192.04 eV; see Figure 7b for an illustration and Figure 3b for the XANES peak at Eₚ (B) = 192.0 eV. Such a configuration essentially requires a SiO₂ shell with small x values as an environment to exist because the B bridge separates the two 1 nm Si atoms by ≥3.66 Å. This environment exists not quite as near to the pure Si NC surface as found for the ≡Si—B(—O—Si≡₂)ₓDB—Si≡ configuration, therefore near the surface of the pure Si NC within the SiO₂ shell. Another indicator for this location is given by the immense growth of the associated XANES peak with NC diameter, viz., going from hardly noticeable for sample 3 nm-H to considerable intensity for sample 6 nm-H, where the SiO₂ shell is significantly thicker.\[^{33,34}\]

Another XANES peak related to B in a SiO₂ environment is given by B on a Si site in SiO₂, a configuration which is shown in Figure 4c. Its E₁₅₁₁ (B) value of 193.16 eV is indistinguishable in experiment from the ≡Si—B(—O—Si≡₂)ₓDB—Si≡ configuration discussed earlier. To add, both configurations share the SiO₂ phase as their optimum environment.

The last configuration of B involving O only in its proximity is shown in Figure 7c, displaying the only case where B does not bond to the local lattice. Appearing strange at first glance, the interstitial O=O—OH molecule has its origin in the presence of residual O in the proximity of B at high temperatures, where B acts as a strong reduction agent. For comparison, forming of boron nitride (BN) along 2B + N₂ → 2BN + 510 kJ mol⁻¹ (4.14 eV/B atom) takes place at 900 °C, whereas the equivalent reaction of Si with N₂ does not occur below ≈1200 °C.\[^{26}\]

During the anneal at 1100 °C, BOOH can form if reactive/residual O is present in the sample, which is likely the case for SiOₓ precursor layers where Si NCs are formed. A corresponding XANES peak to E₁₅₁₁ (B) = 193.48 eV appears to be situated at Eₚ (B) = 193.65 eV, whereby its dependence on NC size plainly shows that the location of BOOH is within the SiO₂ shell surrounding the Si NCs as is the case in previous O-related B configurations. Once again, the broad peak of sample 3 nm-H does not allow us to discriminate between signals originating from BOOH versus B on an Si site in SiO₂ versus the ≡Si—B(—O—Si≡₂)ₓDB—Si≡ configuration. The presence of N₂ during annealing along with reactive/residual O might also lead to a formation of BN, although the formation of O-related B configurations outruns BN formation in terms of reactivity and reaction enthalpy by far.\[^{26}\]

4. Conclusion

B-doped, SiO₂-embedded Si nanocrystal samples and respective SiO₂: B and bulk-Si:B reference samples were studied by XANES at the B K-edge and by real-space DFT. The XANES spectra of the

![Figure 7](https://example.com/figure7.png)

**Figure 7.** Relevant regions of DFT approximants showing B in H-terminated 2.2 nm Si NC (nominal Si₂₈₆ core) with participation of O atoms. a) We start with the same NC approximant as in Figure 5c, whereby two of the three direct bonds of B to its 1 nm Si are bridged by O=Si₂₈₆(—B(—O—)ₓH₁₄₄), 1 nm Si DB. b) The nominal Si₂₈₆H₁₄₄ approximant with one Si—Si bond broken up and bridged by >B—OH=Si₂₈₆H₁₄₄ with Si=Si-(OH)—Si substituted for Si—Si. c) At high temperatures, B can form BOOH (O=O—OH) as a precursor to boric acid (B(OH)₃, sum formula H₃BO₃) being created under O deficiency in Si, as shown here in the nominal Si₂₈₆H₁₄₄ NC. As for atom colors, we refer the reader to Figure 4.
Si NC samples with low average B concentration (0.24–0.31 at%) feature mainly one peak at 194.25 eV, which is attributed to B on a Si site in SiO₂ (structure 4a). Increasing the average B concentration to 0.66–0.88 at% leads to the occurrence of B dimers with Si bridge bonds and (un)passivated Si dangling bonds with XANES peaks at 192.00–192.15 eV (structures 5a–c). Also, broad peaks at 200 and 204 eV occur, which are typical features of borosilicate glasses and B zeolites. Finally, two features on the shoulder of the 194.25 eV peak occur at 193.65 and 193.4 eV, which can be attributed to either BOOH (structure 7c) or a B atom where two of the three Si–B bonds to the 1nn Si are bridged by O (structure 7a). All of these high-[B] features demonstrate that an excessive amount of B does not enrich the B concentration in the Si NCs but forms new configurations that are inherently incompatible with electronic doping. Therefore, it is concluded that B in the Si NC/SiO₂ system is predominantly incorporated in the SiO₂ transition shell at the Si/SiO₂ interface with a partial O coordination as well as in the SiO₂ matrix surrounding the NCs, provided the B concentration remains below the threshold to hyperdoping and the formation of Si borides. The findings of our combined XANES-DFT study are well in accordance with recent results about the location of B in Si NCs obtained from APT,[3,12,13] current–voltage measurements,[19] and high-energy X-ray diffraction.[10]

The mounting evidence that electronic B doping of low nanoscale Si nanostructures fails for a multitude of fundamental physical reasons requires the investigation of alternative strategies that do not rely on classical impurity doping. Acceptor modulation doping of SiO₂[25,35] and the electronic structure shift in low nanoscale intrinsic Si by ultrathin coating with SiO₂ versus Si₃N₄[36,37] are two such strategies which allow us to induce a strong electron or hole dominance in analogy to conventional doping without its detrimental side effects and associated size limits.

5. Experimental Section

Sample Fabrication: Multiple bilayer stacks of B-doped Si-rich oxide (SRO:B, with stoichiometry parameter SiO₁₋ₓCₓ) and SiO₂ were deposited on wet-chemically cleaned Si substrates (n-type, 5–250 Ω cm) by plasma-enhanced chemical vapor deposition (PECVD) using SiH₄, 10% B₂H₆/SiH₂Cl₂, and Ar.[38] At first, a 30 nm thick SiO₂ buffer layer was deposited on the wafers to prevent diffusion of B dopants into the lowly phosphorus-doped substrates. Note that a diffusion length of only ≈7 nm for B in SiO₂ was reported before for a comparable sample system and process conditions.[9] A total of 50 bilayers of either 3 nm or 6 nm SRO:B and 3 nm SiO₂ were deposited and capped with a 10 nm thick SiO₂ layer. Two different nominal B concentrations in the SRO:B layer were used, 1.32 and 0.47 at%.[39] SiO₂:B reference samples were fabricated with 225 nm thickness, 0.8 and 0.25 at% B concentration, and the same buffer/capping layers. All B concentrations were determined by molecular Cs⁺ SIMS (MCs⁺-SIMS).[15] After deposition, the samples were annealed in a quartz furnace at 1100 °C for 1 h in a high-purity N₂ atmosphere and subsequently at 450 °C for 1 h in pure H₂ to grow and crystallize the Si NCs and to passivate Si NC/SiO₂ interface defects, respectively.[39]

For comparison to ionized B dopants in bulk Si, a reference sample was prepared by ion implantation. A boron dose of 1 × 10¹⁶ cm⁻² was implanted at 20 keV (resulting in a projected range of ≈80 nm where the initial B concentration reached ≈10¹⁰ cm⁻²) and subsequently annealed at 1000 °C for 45 min. After annealing the sample was etched in a hydrofluoric acid (HF) solution to remove any silicon oxide grown in the furnace. According to four-point-probe measurements, a sheet resistance of 19Ω°⁻¹ was obtained. No additional HF dip was performed directly before XANES measurements so that the sample was passivated with a native oxide.

Measurements: PL was measured using a LN₂-cooled CCD camera attached to a single grating monochromator with excitation by a HeCd laser (325 nm, ≈3 mW cm⁻²). Electrochemical capacitance–voltage (ECV) profiling was conducted using a CV2P1 system from WEP. XANES and EXAFS measurements were performed using the B K-edge was measured at IQMT’s soft X-ray analytics facility WERA at the KIT synchrotron light source KARA, Karlsruhe, Germany. The photon-energy resolution was set to 100 meV, and a windowless four-element silicon drift detector was used to record XANES spectra in PEF detection mode, with the detection window set around the B K fluorescence. A NiO reference sample measured in parallel to all B K spectra provided an energy reference signal (Ni L edge in fourth order). It enables correction of a “global shift,” which amounted to ≈120 meV. A number of individual spectra showed a maximum shift of (+40, −80) meV relative to this “global shift.” These relative shifts for individual spectra were too small to influence the results, where energy distances of 300 meV appear important, and thus were not corrected for here. The B 1s PFY-XANES sampling depth in Si and SiO₂ was ≈30 and 50 nm, respectively. A strong spectral background (that may include scattering effects) was removed in the energy windows shown, and the spectra were corrected for variations in the incoming intensity, I₀. The fluorescence efficiency is low at B 1s, so for some samples, many individual spectra were averaged to increase the signal-to-noise (S/N) ratio.

DFT Calculations: Real-space calculations were conducted with a molecular orbital basis set (MO-BS) and Hartree-Fock (HF)/DFT methods, using the Gaussian09 program package[40] with the GaussView program[41] for visualization. Initially, the MO-BS wavefunction ensemble was tested and optimized for describing the energy minimum of the system (variational principle; stable – opt) with the HF method.[42,43] Exact exchange interaction inherent to HF is crucial in obtaining accurate bond geometries.[36] As MO-BS, the Gaussian type 3-21G MO-BS (HF/3-21G) was used.[42] The HF/3-21G route was used for the structural optimization of the approximants to arrive at their most stable configuration (maximum integral over all bond energies); root mean square (RMS) and peak force convergence limits were 8 and 12 meVÅ⁻¹, (300 and 450 μPa Å⁻¹), respectively. Optimized geometries were used to calculate their electronic structure by testing and optimizing the MO-BS wavefunction ensemble with the nonlocal B3LYP hybrid DFT method[44,45] and the Gaussian type 6-31G(d) MO-BS, which contains d-polarization functions (B3LYP/6-31G(d)) for all chemical elements. For all calculations, tight convergence criteria were set to the self-consistent field routine and no symmetry constraints to MOs were applied. Ultrafine integration grids were used throughout. The supporting information to König et al.[36,50] contains accuracy assessments.

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