Real-Space Mapping of Surface-Oxygen Defect States in Photovoltaic Materials Using Low-Voltage Scanning Ultrafast Electron Microscopy

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ABSTRACT

Ultrathin layers of native oxides on the surface of photovoltaic materials may act as very efficient carrier trapping/recombination centers, thus significantly affecting their interfacial photophysical properties. How ultrathin oxide layers affect the surface and interface carrier dynamics cannot be selectively accessed by conventional ultrafast transient spectroscopic methods due to the deep penetration depth (10s-100s nm) of the pump/probe laser pulses. Herein, scanning ultrafast electron microscopy (S-UEM) at a low voltage of 1 keV electrons was recently developed at KAUST to selectively map the ultrafast charge carrier dynamics of a few layers on the top surfaces of photovoltaic materials. Unlike high voltage 30 keV experiments, at 1 keV, the depth of detected secondary electrons (SEs) underneath the surface is significantly reduced five times, thus making it possible to visualize the dynamics of charge carrier from the uppermost surface of photoactive layers. More specifically, this new approach has been employed to explore and decipher the tremendous impact of native oxide layers and oxygen defect states on charge carrier dynamics in space and time simultaneously at sub-nm levels on several photovoltaic material surfaces, including Si, GaAs, CdTe and CdZnTe single crystals. Interestingly, the contrast in the SEs time-resolved difference images switched from a dark “energy-loss mechanism” to a bright “energy-gain mechanism” only by removing the layers of surface oxides. Moreover, the charge carrier recombination was estimated and found to be dramatically affected by the native oxide layers. The density functional theory (DFT) calculations demonstrate that the work function of oxygen-terminated Si surface also increases slightly upon optical excitation and makes for less SEs intensity, providing another piece of evidence that the origin of the dark contrast observed on these material surfaces should be assigned to the surface oxide formation, mainly oxygen defect states in the bandgap, and/or work function increment. Our findings provide a new method and pave the way for evaluating the effect of surface morphology and defects, including but not limited to native oxide, on charge carrier dynamics and interfacial properties of photovoltaic absorber layers.

KEYWORDS: photovoltaic materials, scanning ultrafast electron microscopy, native oxides, surface dynamics, energy-loss mechanism, energy-gain mechanism, density functional theory

INTRODUCTION

Solar cells with remarkable power conversion efficiency, reduced cost and minimal environmental effects are a promising candidate to replace fossil energy.¹ Although numerous types of solar cells, such as organic,² inorganic,³ plasmonic⁴ and perovskite,⁵⁻⁶ with excellent optical properties have been reported, the majority of commercialized solar cells with high power conversion efficiency are based on crystalline Si,⁷ GaAs⁸ and CdTe,⁹ and they may also establish the basis of next-generation commercialized solar cells.¹⁰ In this case, a profound understanding of their ultrafast photophysical processes, such as charge-carrier generation, trapping, charge separation and collection in the top surface layers and device interface, is essential to further develop an ideal interface for highly efficient photovoltaic devices. In this regime, surface and interfacial engineering aims at promoting the collection and minimizing the recombination of charge carriers are crucial for high-efficiency solar cells.¹¹ Among these, surface passivation¹² and surface activation¹³ are the most popular strategies in solar research communities. However, how to characterize and precisely evaluate the effect of ultrathin surface engineering on the photogenerated charge carrier dynamics before and after surface and interface treatment remain largely unsolved.

Methodologies such as laser pump-X ray probe,¹⁴ lock-in carrierography¹⁵ and transient photocurrent/photovoltage measurements¹⁶ have been employed to visualize the surface and interface properties of photovoltaic materials but due to the large penetration depth of the pump light of these technologies, the obtained transient data mainly reflect the bulk properties.¹⁷ Therefore, new methodologies that are capable of real-space mapping of trap densities, defects, carrier diffusion and recombination processes selectively at the uppermost surfaces (from angstroms to a few nanometers beneath the surface) are urgently required.

Scanning ultrafast electron microscopy (S-UEM) is a state-of-the-art apparatus that enables the visualization of surface charge carrier dynamics in real time and space simultaneously by combination of a femtosecond laser (Clark-MXR) and modified scanning electron microscopy.¹⁸⁻

²⁰ The schematic description of the S-UEM setup is shown in **Figure 1A**. An optical laser pulse with a wavelength of 515 nm is set as the pump light to optically excite the surface of the photovoltaic materials. The probe pulse in this case is not a continuous electron beam that is generated by thermal heating of the Schottky field-emitter tip, as in conventional scanning electronic microscopy (SEM); instead, a delayed UV pulse centered at 343 nm is employed to excite the cooled aforementioned tip to generate doses of pulsed primary electrons (PEs) that interact with the surface of the specimen to generate secondary electrons (SEs) signals. Detailed information on the S-UEM setup was published elsewhere.¹⁸ In this S-UEM experiment, the optical excitation promotes a fraction of the valence band (VB) electrons to the conduction band (CB).¹⁸ Upon interaction with the energetic PEs beam, the promoted photogenerated electrons are energetically easier to migrate above the vacuum level, increasing the probability of SEs emission (i.e., energy gain mechanism). Consequently, the contrast should be bright in the SEs time-resolved difference images, as detected for the clean surfaces of CdSe and Si.¹⁸⁻¹⁹ However, a dark image contrast is also observed after surface excitation in a number of photovoltaic materials, including GaAs,²¹ CIGSe²² and InGaN NWs²²⁻²³. This dark contrast suggests that the density of SEs decreases after optical excitation (fewer SEs escape from the material surface). It should be noted that the reason behind this energy-loss mechanism has been assigned to scattering processes between the SEs and the photogenerated carriers in the CB, as the effective scattering cross section is larger for the SEs generated in the CB than in the VB.^{18,21} The natures of these interactions were explained to be electron-phonon, electron-electron, and inelastic scattering processes.^{21,24} It should

be noted that all mentioned interactions and materials (CdSe, Si, GaAs, CIGSe and InGaN) have been previously measured at an accelerating voltage of 30 keV.^{18-19, 22, 25-26}

Figure 1. (A) Schematic of the S-UEM experiments at 1 keV before and after surface cleaning; (B) SEs images at different time delays for as-received; and (C) fresh HF- freshly etched Si (100) upon 515-nm excitation. The dashed ellipse indicates the laser footprint on the material surface.

Working with different acceleration voltages not only affects the penetration depth of the primary electron beam but also the energy and depth of the generated SEs near the surface. This is because the penetration depth of pulsed PEs is controlled by the electron energy and the inelastic scattering processes with the atomic electrons, as well as by deflection due to nuclear collisions.²⁷ The shape of the energy distribution of SEs can be described by **equation (1)**.²⁸ It shows that any factors that affect the work function or Fermi energy will result in a change in the energy distribution of SEs.

$$\frac{dN}{dE} = k (E - E_F - \phi) / (E - E_F)^4 \quad (1)$$

where E is the energy of SEs, k is the material constant, ϕ is the work function and E_F is the Fermi energy.

Several theoretical studies estimated the escape depth of SE of various materials at different applied voltages,²⁹⁻³⁴ showing that increasing the acceleration voltage of the primary electrons would result in the SEs with higher energy that escape from deeper layers. Herein, we select the following approximations to compare between conditions under low and high applied voltages on the escape depth of SEs. With integration of the energy distribution, the total cross-section of SEs emission and the transmission fractions of SEs and some rational approximations, the most probable escape depth of SEs (d_{esc}) can be obtained as shown in **equations (2) and (3)**:

$$d_{esc(nm)} = 0.267 \frac{AI}{\rho Z^{2/3}} \quad (5 \text{ keV} \leq E_0 \leq 1 \text{ MeV}) \quad (2)$$

$$d_{esc(nm)} = 0.0534 \frac{AI}{\rho Z^{2/3}} \quad (0 \text{ keV} < E_0 < 5 \text{ keV}) \quad (3)$$

where A is the atomic weight (g/mol), ρ is the density (g/cm³), Z is the atomic number, I is the first ionization potential (eV) of metal specimens (or electron affinity (eV) for semiconductor

specimens), and E_0 is the incident energy (i.e. acceleration voltage). A detailed description of the derivation of these equations is shown in the **Supporting Information**.

We can facilely calculate the escape depth of SEs detected in real space and time domains by these equations, and we found that the majority of SEs come from the uppermost several nanometers (~ 5 nm) and ~ 1 nm beneath the surface at 30 keV and 1.0 keV, respectively. In addition, the density of SEs increases with the applied voltage, ca. five times, see **Figure S1**. The d_{esc} of SEs in the photoactive materials investigated in this study are shown in **Table S1**, and these values are in accordance with previously reported theoretical results^{28, 35} and experimental data³⁶.

The surface oxides grow naturally on Si,³⁷ GaAs,³⁸ CdTe,³⁹ and CdZnTe⁴⁰ and regrow very rapidly (in minutes) on the freshly cleaned surface of these photoactive materials.⁴¹⁻⁴³ These native oxide layers are ultrathin in nature, usually in the range of sub-nm to a few nm,^{37, 44} but they could play critical roles in controlling the surface charge carrier dynamics, which is still not easy to be fully explored using S-UEM at 30 keV due to the large difference between the thickness of these ultrathin native oxide layers and the escape depth of the SEs. In this work, by employing 1 keV as the applied acceleration voltage, the surface depth from which the SEs are generated and detected from is the uppermost surface (1 nm or even much less) and enables the observation of the exact surface dynamics from the examined materials. Interestingly, we observed both dark and bright contrasts in the same materials (Si, CdZnTe, CdTe, and GaAs) upon optical excitation only by cleaning the surfaces of native oxide layers, as revealed by the XPS results. Thus, we can conclude that, at least in these four materials, the surface oxides are the origin of the energy-loss mechanism (dark contrast), which could be attributed to charge carrier trapping and/or work function changes, as evidenced by DFT calculations (see later section).

Experimental Methods

Silicon crystal with n-type-doped silicon wafers were purchased from Norsun. The surface of crystals was treated as previously reported.¹⁷ Briefly, specimens with 1 cm × 1 cm size were cut by diamond knife from phosphorus doped n-type Si wafers (6 inch, resistivity of 1-5 ohm/cm, with doping level of $\sim 6.5 \times 10^{14} \text{ cm}^{-3}$). These specimens were firstly immersed into hot (90 °C) mixed solution of H₂SO₄ (98%) and H₂O₂ (30%) with volume ratio 3:1 for 10 minutes, subsequently, they were taken out and put into a 5% HF solution for 30 seconds and washed by deionized water for several times. After that, they were dried with ultrapure N₂ and directly transferred into the vacuum chamber of S-UEM apparatus for the time-resolved experiments. During the transfer process, specimens were sealed in portable vacuum desiccator to minimize the effects of oxygen and moisture. CdTe (111) undoped p-type, CdZnTe (111) (Cd_{1-x}Zn_xTe, x = 0.04) p-type and GaAs (110) VGF grown undoped single crystals were obtained from MTI Corporation. The surfaces of the CdTe, CdZnTe, and GaAs crystals were cleaned by argon etching. The samples were first measured as-received by S-UEM at an applied voltage of 1 keV. The XPS measurements and argon (Ar) etching were performed by an Axis Supra DLD spectrometer (Kratos Analytical) equipped with a monochromatic Al K α X-ray source ($h\nu = 1,486.6 \text{ eV}$) operating at 150 W, a multichannel plate, and a delay line detector under a vacuum of $\sim 10^{-9}$ mbar as previously reported.⁴⁵ Based on previous reports, the thickness of native oxide layers on all the selected materials is less than 1 nm when they were exposed to air/ moisture for a short period of time (from hours to days),⁴⁶⁻⁴⁸ furthermore, all the samples that measured as received in the present work were initially stored in sealed boxes which saturated with Ar and after taken out, they were stored in vacuum chamber so the thickness of less than 1 nm can be guaranteed for the oxide layer. All S-UEM experiments reported here were conducted at a repetition rate of 8 MHz to introduce an interval of 125 ns for successive events, making sure full recovery of the system back to the initial state before the next

pulse comes. It should be mentioned that the temporal resolution of S-UEM was measured to be 650 fs, and the focal spot of the excitation pulse is an oval shape with size of 40 μm . Both the photo-irradiated and non-irradiated areas were scanned by pulsed primary electrons to obtain time-resolved SE difference images. These images were obtained through integration of 64 frames and 300 ns dwell time. The Density functional theory (DFT) calculations were carried out at GGA/PBE level using PAW method in Vienna Ab initio simulation package (VASP).⁴⁹⁻⁵⁰ The non-passivated, H-terminated, and O-terminated Si(100)-c(4×2) slabs were built and optimized to investigate the electronic properties. The hybrid functional Heyd-Scuseria-Ernzerhof (HSE) (an exact exchange fraction = 0.18) was used to calculate the vacuum levels, Fermi levels and work functions of the studied slabs.

RESULTS AND DISCUSSIONS

The concept of our time-resolved imaging approach at low voltage to map the surface carrier dynamics in four different photovoltaic materials before and after surface treatments is shown in **Figure 1A**. As can be clearly seen in the XPS data (see later section), etching of the surface removes the native oxide layers and subsequently surface oxygen defects. For as-received Si (100), i.e., before any cleaning process, the SEs time-resolved difference images showed a dark contrast upon optical excitation (**Figure 1B**). In sharp contrast, after the HF etching procedure and removing the native oxide layers, the SEs time-resolved difference images showed a bright contrast (**Figure 1 C**) instead of the dark contrast seen on the same samples before etching. This observation provides new physical insights into the significant influence of the oxide layers on the SEs generation mechanism and charge carrier dynamics on top surfaces. There are two possible explanations for such a contrast change from bright to dark due to the formation of native oxide

layers. The first explanation is the change in the work function of the surface after oxide layer formation. It should be noted here that the work function controls the SEs emission from the surface, as revealed by the aforementioned **equation (1)**. This would mean that any change in the work function directly changes the density of the SEs emission on the surface that could change the sign of the image contrast. However, DFT calculations (**Figure 2A-C**) demonstrate that the H passivation and oxide termination slightly reduce the work function from 5.68 eV to 5.57 eV and 5.32 eV, respectively. We further tune the horizontal Si-O distance within 0.06 Å to mimic the surface structural change upon the optical excitation for the O-passivated Si(100)-c(4 × 2) slab (the calculated total energies and work functions are given in **Figure 2D**). We find that the surface oxygen atoms are inclined to move towards top-surface Si atoms upon optical excitation to achieve a more stable slab with a small total energy change; and the work function increases from 5.32 to 5.60 eV by increasing or decreasing the horizontal Si-O distance. In this case, the less SEs intensity at Si surface terminated by oxygen atoms will be obtained, which results in darker contrast, comparing to non-terminated surfaces by oxygen and is in accordance with experimental results here. The influence of these change is counteracted due to the background deduct method and difference image mode we adopted. The second possible explanation is that the native oxide layers create photogenerated oxygen defects in the bandgap upon optical excitation, which act as quenching centers for the excited carriers. In this regime, our previous DFT calculations (which can provide direct information on surface charge density properties) indicated that the partially passivated surface by oxygen atoms results in ultrafast carrier trapping on the surface of Si, leading to a significant reduction in the SEs efficiency and charge carrier lifetime.¹⁷ Thus, it is most likely that the oxygen defect formation is mainly responsible for the change in the charge carrier

dynamics and the contrast in the presence of the native oxide layers, while the work function increment may contribute in some extent as well.

Figure 2. Optimized crystal structures and corresponding vacuum levels, Fermi levels and work functions of (A) non-passivated, (B) H-passivated, and (C) O-passivated Si(100)-c(4 × 2) slabs; (D) calculated total energies and work functions of O-passivated Si(100)-c(4 × 2) slabs by tuning the horizontal distance (d) between oxygen atoms and top surface Si atoms from original position

($x = 0$). The total energies of slabs were calculated at the GGA/PBE level and work functions were calculated at the HSE level based on the electronic wavefunctions at the GGA/PBE level.

The SEs kinetics before and after HF etching are shown in **Figure 3A**. The generation of SEs after HF etching occurs at approximately 130 ps, which was previously assigned to the diffusion of photogenerated carriers to surface,¹⁷ while the disappearance of SEs due to charge carrier recombination is much faster at the as-received Si surface than that of at the freshly HF-etched surface of Si. This suggests that some carrier trapping sites at the surface have been removed by HF etching. To investigate the effect of etching on the chemical composition of the Si (100) surface, XPS was performed for samples before and after etching (**Figure 3B**). The main difference in the XPS spectrum after etching is the significant decrease of the SiO₂ peak that appears at a binding energy of 103.5 eV in Si before etching.⁵¹⁻⁵² Note that the slight increase in the oxidation level in the etched Si can be a result of exposure to air during the fast transfer process after etching to the XPS instrument.

Figure 3. (A) Extracted dynamics of the time-resolved SEs intensities of the laser footprint region in the as-received Si before (black) and after HF etching (red). (B) Si 2p XPS spectra of the as-received (black) and HF freshly etched Si (red) with the same intensities.

It is worth noting that the same Si (100) samples were measured at an acceleration voltage of 30 keV before and after HF etching. In both cases, the SEs time-resolved difference images showed a bright contrast but with totally different charge carrier recombination dynamics.¹⁷ In this study, XPS, kelvin probe-force microscopy (KPFM), and DFT calculations along with S-UEM were employed to explore and decipher the role of native oxides as charge trapping centers within the bandgap. The time-resolved SEs difference images and KPFM results prove that the native oxide layer regrows very rapidly after removal by surface treatment, producing unwished surface-trap states that quench the population of photo-excited charge carriers on the sample surface. The difference in the contrast for the same sample (i.e., as-received Si (100) before any treatment) with varying accelerating voltages can be understood in regard to the role of the applied voltage in controlling the surface depth from which the SEs are generated and subsequently detected. Because 30 keV generates SEs from deeper surface layers with higher energy and carrier density than those generated at 1 keV, a bright contrast can be observed despite the presence of few layers of native oxides on the top surface.

Figure 4. SEs difference images as function of time delays for (left) as-received CZT (111) crystal and (right) after Ar etching upon 515-nm excitation. The dashed ellipse indicates the laser footprint on the surface.

Is this phenomena applicable to other photovoltaic materials or affected by the etching and doping methodologies? To answer this question, we carefully examined CdZnTe (CZT) and CdTe with the same orientation as (111) and GaAs (110) samples before and after Ar etching. **Figure 4** shows the SEs time-resolved difference images for CZT before and after Ar etching. The left panel shows the SEs time-resolved difference images before etching, where a clear, dark contrast is observed. After Ar etching (right panel), the contrast becomes bright, confirming the tremendous impact of the native oxide layers on the SEs emission in this material as well. **Figure 5A** shows the SEs kinetics before and after Ar etching for the CZT sample. Similar to the Si case, the generation of SEs before and after Ar etching occurs on a similar time scale of approximately 420 ± 20 ps. On the other hand, the charge carrier recombination is faster for the untreated surface than the treated surface, confirming the role of the defect states due to the layers of oxide in accelerating

carriers to recombine. Examining the surface chemical composition by XPS revealed that the main difference in the CZT sample before and after etching with Ar is the removal of the TeO_2 peaks found at binding energies of 575.9 eV and 586.3 eV,^{53,54} as shown in **Figure 5B**.

Figure 5. (A) Extracted dynamics of the time-resolved SEs intensities of the laser footprint region in the as-received CZT (111) before (black) and after Ar etching (red). (B) Te 3d XPS spectra of the as-received (black) and Ar-etched CZT (red) with the same intensities.

Similarly, contrast changes from dark to bright after Ar etching are observed for CdTe (111) crystals. As shown in **Figure 6A**, the SE kinetics before and after Ar etching for CdTe crystal are different. For instance, the generation of SEs before and after Ar etching occurs at nearly identical time scales (225 ± 5 ps), while the decay of SEs is much faster at the surface before Ar etching, which is very consistent with the observation in the Si and CZT crystals. To confirm the presence of the native oxide layers in this material, XPS experiments were conducted. **Figure 6B** shows the high-resolution XPS of Te 3d, revealing that the main difference in the CdTe sample before and

after etching with Ar is the removal of the TeO_2 peaks found at binding energies of 576.0 eV and 586.4 eV.^{53, 55} This observation again confirms that the oxygen defects created by the formation of native oxide layers are indeed the origin of the dark contrast formation and the fast charge carrier recombination before surface etching. In other words, the results of Si, CdTe and CZT show that the contrast changes from dark to bright with the removal of oxide layers despite the use of different etching and doping methodologies and orientations. This suggests that the surface-oxygen defect states control the surface dynamics, which is probably a universally applicable rule.

Figure 6: (A) The dynamics of the time-resolved SEs intensities of the laser footprint region in the as-received CdTe (111) before (black) and after Ar etching (red). (B) Te 3d XPS spectra of the as-received (black) and Ar-etched CdTe (red) with the same intensities.

Previously, the surface dynamics of the GaAs (110) single crystal was studied by S-UEM at 30 keV, in which a dark contrast was reported and assigned to carrier-carrier scattering due to the strong inelastic scattering processes between the highly penetrated PEs and the photogenerated charges located within the material bulk.²¹ Moreover, it was also claimed in the same work that

the effective cross-section for the scattering of internal SEs with the CB electrons is expected to be much higher than that with VB electrons, which may also lead to dark contrast formation upon optical excitation.

Figure 7: (A) Extracted kinetics of the time-resolved SEs intensities of the laser footprint region in the as-received GaAs (110) before (black) and after Ar etching (red). (B) SEs time-resolved difference images at different time delays for (top) as-received GaAs crystal and (bottom) after Ar etching upon 515-nm excitation. The dashed ellipse indicates the laser footprint on the surface. (C) As 3d XPS spectra of the as-received (black) and Ar-etched GaAs (red) with the same intensities.

Here, we employed 1 keV to access and visualize carrier diffusion and recombination from the bulk to the top surface of the GaAs samples and compared them with the results obtained at 30 keV in references.²¹ The SEs intensities from the GaAs (110) SEs time-resolved difference images are plotted as a function of the time delay between excitation and probe before and after Ar etching, and the extracted SEs kinetics are shown in **Figure 7A&B**. The opposite sign of the SEs difference image contrast is clearly observed before and after etching. To investigate the effect of etching on the chemical composition of the GaAs (110) surface, XPS was also performed for samples before and after etching (**Figure 7C**). The main difference in the XPS spectrum after etching is the

decrease – almost removal – of the As_2O_3 peak that appears at a binding energy of 44.4 eV⁵⁶⁻⁵⁷ before etching. The reason for the slight increase in the oxidation level at the etched GaAs surface is similar to that of the Si sample. On the one hand, at an applied voltage of 1 keV, the carrier-carrier scattering due to strong inelastic scattering processes is decreased dramatically, while a dark contrast is still observed; on the other hand, by removing oxide layers, the contrast changed to bright contrast, proving unambiguously that the inelastic scattering events are not the decisive factor and the change in contrast and carrier dynamics is most likely due to surface oxygen defect states. Being in the regime, the fitting parameters of the charge carrier dynamics of these investigated materials are summarized in **Table S2** in **Supporting Information** for better understanding the impact of native oxides on evolution of photogenerated carriers on materials surfaces.

It is worth mentioning that the SEs signal came out from the material underneath the native oxides layers in as-received samples due to density distribution effect of emitting SEs (see **Figure S1**), other than from the oxide layers directly, which are limited by angstrom thickness. More importantly, the energy of excitation photons utilized in the S-UEM is 2.41 eV is much lower than the bandgaps of oxide layers, such as ~ 6.0 eV for SiO_2 ,^{58,59} ~ 3.5 eV for TeO_2 ,⁶⁰ and ~ 6.0 eV for As_2O_3 ⁶¹. This would mean that the photo-illumination cannot excite the native oxide layers of the materials used in this study.

CONCLUSIONS

We visualized surface charge carrier dynamics simultaneously in space and time for several photovoltaic materials, including Si, GaAs, CdTe and CdZnTe, with and without native oxide layers using S-UEM at an accelerating voltage of 1 keV. Interestingly, we observed that the contrast in the SEs time-resolved difference images changes from dark “energy-loss mechanism”

to bright “energy-gain mechanism”, and the SEs kinetics also changed dramatically only by removing the native oxide layers. These results could not have been reached previously using high applied voltages that penetrate much deeper and result in high density SEs with higher energy that come from relatively deeper areas below the oxide layers. This suggests that the origin of dark contrast observed in some materials at positive time delays should be ascribed to carrier trapping by surface oxides and/or increment of work function due to oxygen termination, not other scattering events as previously reported. Our findings provide a sensitive method of choice for evaluating the nature of charge recombination processes, and pave the way for evaluating the effect of surface morphology and defects such as native oxides on surface dynamics and interfacial properties of photovoltaic absorber layers.

ASSOCIATED CONTENT

Supporting Information. The Supporting Information is available free of charge on the ACS Publications website. Calculation methodologies of most probable escape depth.

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Notes

Any additional relevant notes should be placed here.

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