SUPPLEMENTARY INFORMATION

METHODS

All experiments and analyses were done at the University of Lausanne. A large (>3x2 cm) single crystal of quartz from Brazil (unknown exact origin, likely from Minas Gerais) was cut into slabs perpendicular to the crystallographic c-axis. This was then mounted in acrylic and polished using Al₂O₃ slurry followed by 10, 6, 3, 1, ¼ μm diamond paste, then colloidal silica suspension on a vibrating lap, then cut into smaller (2-4 mm pieces). The polishing was done prior to cutting in order to limit any potential development of edge topography. LA-ICP-MS analysis of the quartz (Australian Scientific Instruments RESOlution 193nm ArF excimer laser coupled to a Thermo Element XR sector field mass spectrometer) gave 41±14 wt. ppm Al, 0.26±0.16 wt. ppm Ti (2 s.d.) and Fe below detection limit.

Two different kinds of Ti source were used, both broadly describable as TiO₂-doped SiO₂, but synthesised in different ways. For the first (herein ‘solid source powder’), TiO₂ (anatase) and SiO₂ (quartz) powder (Sigma-Aldrich) were weighed in various proportions (100, 500, 1000, 5000 and 10000 wt. ppm Ti, and pure TiO₂) then mixed under acetone in an agate mortar. The mixes were dried, pressed into pellets, then sintered at 1300 °C in a box furnace (Borel) for 48 h and re-ground (Fig. S 1). In the main text, this is described as TiO₂-SiO₂, because, given the diffusivities in this study, it is unreasonable to assume that the mix fully sintered so there may be two phases present in the powder (even if not identifiable by Raman Spectroscopy).
Fig. S 1: SEM images (BSE) of some powder sources used in this study. Powders were pressed onto a piece of carbon tape, carbon coated, then imaged. (a) Solid-source powder with nominally 100 wt. ppm Ti, prior to sintering (i.e. anatase plus quartz). (b) As (a), following sintering, and light grinding. Grain sizes from <10 µm to ~200 µm are present. (c) Solgel-source powder with nominal Ti of 100 wt. ppm, following sintering. (d) As (c), for the powder with nominal Ti of 1000 wt. ppm.

Following sintering, all SiO₂ was cristobalite (Raman analysis, see Fig. S 2). LA-ICP-MS analyses of the powders post sintering gave actual Ti contents (all wt. ppm with 2s.d.):

180±8; 496±18; 951±40; 4100±139; 8672±298, Al contents: 3132±157; 3026±98; 3015±97; 3061±99; 3153±99 and Fe contents: 407±17; 416±15; 399±15; 399±14; 406±17 for the 100, 500, 1000, 5000 and 10000 wt. ppm (nominal Ti) mixes, respectively.

Fig. S 2: Raman spectra of crystals and powders used in this study. Also given are the identifiable phases in each spectrum. (a) quartz single crystal starting material. (b) quartz single crystal following experiment. (c) solgel 1 (nominal Ti = 100 wt. ppm). (d) solgel 2 (nominal Ti = 500 wt. ppm). (e) solgel 3 (nominal Ti = 1000 wt. ppm). (f) quartz (untreated, Sigma-Aldrich). (g) anatase (untreated, Sigma-Aldrich). (h) rutile (formed by annealing anatase from (g)) at 1000 ⁰C). (i) solid-source powder, nominal 10000 wt. ppm Ti, before annealing. (j) as (i), following annealing. (k) solid-source powder, nominal 100 wt. ppm Ti, before annealing. (l) as (k), following annealing.
Preliminary experiments, where we attempted to recreate the experimental design of Cherniak et al. (2007) showed considerable interface degradation when using any powders with more than several hundred ppm Ti (Fig. S 3), so the lowest Ti (~100 wt. ppm) mix was used for all solid-source experiments.

Fig. S 3: White light interferometry image of two quartz crystals post-experiment, plus representative transects showing surface topography. (a) shows a crystal annealed with rutile powder for 1 hour at 1200 °C, showing several hundred nanometers of roughness (image width = 2 mm). (b) shows a crystal annealed with a Ti-doped SiO2 powder (experiment TiQ8, 1200 °C, 3 hours) containing nominal Ti of 100 wt. ppm, with considerably less topography than in (a). Note also the SIMS pit in the lower right quadrant (image width = 2.5 mm). We also note, for comparison, that the normal diameter of a beam used for Rutherford backscattering spectroscopy is considerably larger than the size of a SIMS pit – generally on the order of 0.5-2 mm.

The second source (herein ‘sol-gel source powder’) was made using the sol-gel method (Luth and Ingamell, 1965), whereby the Ti-doped SiO2 was precipitated from a solution. For this, ammonium titanyl oxalate monohydrate crystals were dissolved in weak nitric acid, then mixed with tetraethyl orthosilicate and ethanol in a Teflon beaker. These were then placed on a hotplate, with a magnetic stirrer, and slowly dried. The residual solid was then devolatilised
in a Pt crucible over a Bunsen burner, then left in a box furnace at 1300 °C for 12 hr, after which a fine white powder remained. Three mixes were made, with nominal contents of 100, 500 and 1000 wt. ppm Ti. The actual contents (LA-ICP-MS data) were Ti: 115±6; 442±17; 813±34, Al: 5.1±1.4; 18.4±9; 4.4±1, Fe: 16.6±2.9; 82.8±7.6; 13.4±1.8, respectively. All SiO₂ was cristobalite, with a porous texture post-sintering (Fig. S 1, Fig. S 2).

A polished quartz piece was packed into the Ti-doped SiO₂ powder (either solid-source or sol-gel) on a Pt lid, placed into a Borel box furnace, then removed and air-cooled. Temperatures were monitored and controlled using type B (Pt-Rh) thermocouples, with an estimated accuracy of ±2 °C. The secondary (monitoring) thermocouple was assumed to give the correct temperature, as the charge was placed within a few centimetres of the hot junction. The uncertainty on temperature was not included when fitting the Arrhenius relationship, given that it is considerably smaller than the uncertainties associated with diffusion coefficients. Following experiments, the crystals were recovered, cleaned ultrasonically in acetone, isopropanol and ethanol, then mounted in indium and Au coated (35 nm) for SIMS analyses. A single piece (<1 mm) of polished quartz standard (Audétat et al., 2015) was also pressed into each mount.

Experimental products were analyzed via depth profiling on the Cameca 1280-HR SIMS housed in the SwissSIMS facility. Two different analytical protocols were employed, one for experiments conducted using the solid source powder, and another for those employing the sol-gel source powder. This change in analytical protocol was not by intentional design, but was necessary simply because the duoplasmatron source on the Cameca 1280-HR housed at the SwissSIMS facility was replaced by a Hyperion II RF plasma source after our first round of experiments had been analysed; we only decided to cross-check our results using a sol-gel source powder after the Hyperion had been installed.
Samples analyzed using the duoplasmatron source (solid source powder experiments) were held at 5 kV, and were sputtered with a $^{16}$O$_2^-$ primary ion beam (PIB) accelerated at -13 kV. The PIB was run at a beam current of ~25 nA and focused to a ~25 µm spot, which was then rastered over a 250 x 250 µm area. In order to ensure that only ions originating from the center 50 x 50 µm of the rastered area entered the mass spectrometer, we employed both a physical field aperture and electronic gating. This was done in order to avoid ion contributions from the crater walls, so that our measurements had a uniform sampling depth throughout the analyses. Analyses included measurements of $^7$Li, $^{23}$Na, $^{27}$Al, $^{28}$Si, $^{48}$Ti and $^{56}$Fe. All species were counted for 5 s except for $^{28}$Si and $^{48}$Ti, which were counted for 2 and 20 s, respectively. In order to compensate for charging throughout the analyses, we applied an energy re-centering routine using $^{28}$Si at the beginning of each cycle. This routine varies the voltage at which the sample is held, such that the total counts of $^{28}$Si are maximized.

Samples analyzed using the Hyperion II RF plasma source (sol-gel source powder experiments) were held at 10 kV, and were sputtered with a $^{16}$O$_2^-$ primary ion beam (PIB) accelerated at -13 kV. As with the previous routine, analyses included measurements of $^7$Li, $^{23}$Na, $^{27}$Al, $^{28}$Si and $^{48}$Ti. All species were counted for 5 s except for $^{28}$Si and $^{48}$Ti, which were counted for 2 and 20 s, respectively. The PIB was run at a beam current of ~10 nA and focused to a ~10 µm spot, which was then rastered over a 150 x 150 µm area. Field aperture and electronic gating were also used in this routine, but they were setup so that only ions originating from the center 25 x 25 µm area entered the mass spectrometer. The raster area and beam current were reduced in this analytical protocol, due to the fact that the Hyperion source produces a far greater beam current density than that produced by the duoplasmatron, and therefore can more easily produce sample charging. The same charge-compensation protocol was employed in analyses using the Hyperion source as those using the
duoplasmatron. The average energy recentering for these two analytical setups were 0.21 and 0.12 V, respectively. This small change relative to the initial sample voltage (10 kV) indicates that charging was negligible in both analytical routines.

Data from $^7$Li and $^{23}$Na could not be used following either analytical routine, mainly as they both suffered from considerable surface contamination (we found that the Li and Na analyses were the same in unannealed (blank) samples as in experimental samples) and secondly because the SIMS was tuned for maximum sensitivity on Ti.

To assign a spatial dimension to the data, the depth and associated uncertainty of each SIMS pit were determined using white light interferometry (e.g. Wyant (2002)). This was done using a Bruker Contour-GT microscope, with data processed using Bruker Vision64 software. This was done by measuring height versus distance profiles from each SIMS crater (examples shown in Fig. S 4), then extracting the mean and uncertainty of the relative position (relative to an arbitrary zero) of both the crater floor and surrounding sample surface. Only the central ~50 and ~25 µm of the crater was considered for craters made with the duoplasmatron and Hyperion sources, respectively. Then, the position of the crater floor was subtracted from the surface position, and uncertainties propagated by summation in quadrature.
Fig. S 4 Examples of profiles across SIMS depth-profiling craters measured using white-light interferometry. Profiles are from samples: (a) 08_TiQn6; (b) 02_TiQ3; (c) 27_TiQ2; (d) 15_TiQ8; (e) 06_TiQ7; (f) 18_TiQ9. (a), (b), (c) and (d) are examples of relatively poor-quality (rough) pits, (e) and (f) show flat-bottomed, high-quality pits.

The Au-crystal interface in the SIMS data was firstly located using a large increase in Si counts (2-3 orders of magnitude increase). This was assigned to a distance of zero nm, and the rest of the data points assigned distances assuming firstly a constant sputtering rate, and secondly that the Au coat was exactly 35 nm thick. All elements were then normalised to Si. These were then all converted into weight parts per million, using a single value of (Ti/Si)/ppm and (Al/Si)/ppm for the standard for each analytical session. It is important to note that this has no effect on the resulting diffusion coefficients – these could also be determined simply using Ti/Si or Al/Si without conversion to wt. ppm.

The time-series experiments show that whilst the diffusion coefficient appears to be time-independent, the interface concentration often does not (see Table S 1). Non-constant interface conditions in diffusion experiments generally occur where there is a finite amount of the diffusant in the system (Chakraborty et al., 1994). In the present study, it may be that the interface concentration changes over time, as Ti is supplied to the quartz crystal via grain-boundary diffusion in the powder source. Regardless, the diffusion length scale should not be
considerably affected (discussed and modelled below), assuming that there is no concentration-dependence of diffusion, which should not be expected in such a system. To account for this, two different methods were used to extract diffusion coefficients for Ti. The uncertainties from each, defined using the main diagonal of the covariance matrix, were concatenated, to give a total uncertainty for each profile (taking the maximum and minimum values). This also incorporates the uncertainty associated with the measured pit depth, which incorporates uncertainty on the non-analysed surface and the floor of the pit, as described above.

Firstly, concentration versus distance data were fitted using non-linear least-squares regression to the commonly-used solution to Fick’s second law for one-dimensional, concentration-independent diffusion, with constant boundary conditions in a semi-infinite medium (Crank, 1975).

\[ C(x,t) = (C_1 - C_0) \times \text{erfc} \left( \frac{x}{2\sqrt{Dt}} \right) + C_0 \]  

(1)

C is concentration, x is distance (m), t is time (seconds), C₁ is the concentration at the interface (crystal-powder contact), C₀ is the background concentration in the crystal and D is the diffusion coefficient (m²s⁻¹). This was done in two stages. Firstly, the whole profile (excluding the data from the transient zone) was fitted, to extract an approximate length-scale for diffusion. Then, the data from the part of the profile closer to the interface than \(2\sqrt{Dt}\) was removed, and the rest was refitted, to give a final D with uncertainties. This effectively fits only the low concentration tail of the profile.

Secondly, the profiles were fitted to another solution to Fick’s second law (Crank, 1975), which assumes a linear change in interface concentration.
\[ C(x,t) = (C_{1,t=0} - C_0) \left( \text{erfc} \left( \frac{x}{\sqrt{4Dt}} \right) + 4kt \cdot i^2 \text{erfc} \left( \frac{x}{\sqrt{4Dt}} \right) \right) + C_0 \]  

where:

\[ i^2 \text{erfc}(z) = \frac{1}{2} \left( \text{erfc}(z) - 2z \left( \frac{1}{\sqrt{\pi}} e^{-z^2} - z \cdot \text{erfc}(z) \right) \right) \]

The \( k \) term describes the rate of change of the interface concentration, as a function of time, and was allowed to be either positive or negative in the fitting. Such a method has previously been used to fit experimental data from systems where the interface concentration has depleted over time (e.g. \(^{26}\text{Mg} \) diffusion in olivine (Chakraborty et al., 1994)).

**TITANIUM DIFFUSION RESULTS**

Ti diffusion profiles (Fig. S 4) were successfully fit to both equations. The multiple time series were generally successful in terms of diffusion coefficients for Ti (Fig. S 5), but not in terms of the Ti interface concentrations, measured against a quartz standard (Fig. S 6). The variability is likely due to variable contact between the powder source and crystal, both within (across the sample surface), and between experiments. The diffusivities determined with the sol-gel source were slightly lower, albeit within uncertainty of, those determined using the solid-source powder, with the caveat that the latter were analysed following the replacement of the oxygen source in the SIMS, as described above.

The diffusivity data were fitted, with weights, to the general isobaric Arrhenius relationship:

\[ \log_{10} D = \log_{10} D_0 (m^2 s^{-1}) + \left( \frac{-E_a (kJmol^{-1})}{2.303RT} \right) \]

where \( E_a \) is the activation energy for diffusion, \( \log_{10} D_0 \) is \( \log_{10} D \) at \( 1/T = 0 \), \( T \) is the temperature in Kelvins and \( R \) is the universal gas constant. Including all data, \( E_a = 311 \pm 12 \) kJmol\(^{-1} \) and \( \log_{10} D_0 = -8.3 \pm 0.4 \) m\(^2\)s\(^{-1} \), with a reduced chi-squared (\( \chi^2_p \)) of ~1.5.
As discussed in the main text, known analytical artefacts from Rutherford Backscattering Spectroscopy and SIMS cannot explain the discrepancy between these data and those of Cherniak et al., 2007. Whilst surface roughness could potentially be an issue, there is now a growing body of evidence that a single element in a single mineral can have multiple diffusion mechanisms (even without changing valence state), and thus diffusivities. Where proposed, this is generally assumed to relate to one element having multiple potential site occupancies. Some systems where this has been proposed include H in olivine (e.g Padron-Navarta et al., 2014), Li in olivine (Dohmen et al., 2010), Ti in olivine (Cherniak and Liang, 2014; Jollands et al., 2016) and Ar in quartz (Clay et al., 2010). This cannot be definitively ruled out for Ti in quartz – there may be, for example, an interstitial mechanism and a vacancy mechanism for diffusion, which could be expected to have different diffusivities. If one of these mechanisms (i.e. the faster mechanism) only functions at high Ti concentrations, this could explain the differences. This is, of course, highly speculative, and would need to be experimentally tested, or observed in natural systems.
Fig. S 5 Processed Ti diffusion profiles, including fits to the relevant solution to Fick’s second law. Ti (counts per second) have been normalised to Si (counts per second), then converted into weight parts per million Ti using the mean value of the measured standard (Audétat et al., 2015). The data was assigned a spatial dimension using the measured depth of the SIMS pits, then fitted to eqs (1) and (2). Experiments denoted ##_TiQ# are solid-source powder experiments, those denoted ##_TiQn# are sol-gel source powders experiments. See Table S1 for more information. Grey boxes are for visual assistance – these delimit experiments done at the same temperature.
Fig. S 5: Time series data showing diffusion coefficients extracted from experiments done at 1000-1400 °C. Black circles represent data from solid-source powder experiments, and blue circles from sol-gel source powder experiments. Uncertainties are 1 s.d.

Fig. S 6: Ti/Si and Al/Si values measured on the standard (Audétat et al., 2015) over one four day analytical session. (a) Ti/Si, showing the mean and lines representing ± 7.0 %, the uncertainty on the standard value (57±4 wt. ppm). (b) as (a) for Al/Si, with +9.8% (154±15 wt. ppm). Because our data fall within the stated uncertainty of the standard, it is not possible to determine whether the spread in our data relates to heterogeneity in the standard (it is a natural quartz crystal), or some analytical drift.
Fig. S 7: SIMS depth profile measured on a blank (not annealed) crystal, showing the Al/Si and Ti/Si channels. Both elements are slightly heterogenous in the starting material, as expected for natural hydrothermal quartz (Lehmann et al., 2009; Rusk et al., 2008). Note that Ti shows an apparent ~60 nm long profile (likely resulting from ion beam mixing), placing a lower limit on measurable profile lengths using our protocol.

**ALUMINUM DIFFUSION RESULTS**

Whilst not the primary focus of our study, Al diffusion coefficients could also be extracted based on some contamination in the solid-source powder, and much lower contamination in the sol-gel source powder. These data are presented in Fig. S 8, with the data presented on a 10000/T versus log_{10}D plot in Fig. S 9. Fitting to eq. (4) gives E_a=300.2±14.4 kJmol^{-1} and log_{10}D_0=-8.65±0.52 m^2s^{-1} (1 s.d.), again, around 2 orders of magnitude slower than the previous determination of Tailby et al. (2018) (E_a=199±10 kJmol^{-1}, log_{10}D_0=-10.6±0.55 m^2s^{-1}). It is striking that the Al diffusivities of Tailby et al. (2018) and the Ti diffusivities of Cherniak et al. (2007) are similar, and so too are our Al and Ti diffusivities, which might point towards some systematic offset associated with one of the analytical or experimental techniques.

Based on the likely substitution mechanism of Al, which involves monovalent cations (e.g. (Götze et al., 2004; Halliburton et al., 1981; Kats, 1962)), there is a possibility that the monovalent cation content/type (i.e. whether H^+, Li^+, Na^+ etc.) of quartz controls the Al diffusivity, to some extent. This was shown in an internally-consistent dataset by Pankrath and Floerke (1994) and warrants further investigation, although it is beyond our scope. It thus
is possible that neither these diffusion coefficients, nor those of Tailby et al. (2018), are universally applicable, but are only valid for quartz with the same or similar monovalent cation contents as in the experiments.
Fig. S 8: Measured Al diffusion profiles, including fits to eqs (1) and (2). Note that Al content in the experiments conducted using sol-gel source powders (denoted TiQn##) is much lower than the equivalent content in experiments done using the solid-source powders (denoted TiQ##), suggesting that the Al contamination is from the powders rather than the furnace or other labware.
Fig. S 9: Al diffusion coefficients as a function of inverse temperature, plus the data from two previous determinations (Pankrath and Floerke, 1994; Tailby et al., 2018). Pankrath and Floerke (1994) determined values of $k$, which have been converted to $D||c$ using $D||c = k \times 3.25 \times 10^{-20} \text{m}^2\text{s}^{-1}$. This relationship was derived for $\alpha$ quartz, we have assumed that the same relationship can be applied to $\beta$ quartz, hence the Pankrath and Floerke (1994) data from $>573^\circ \text{C}$ should be considered as approximations only.

<table>
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<th>ID</th>
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<th>$T$ (°C)</th>
<th>$t$ (s)</th>
<th>Depth (nm)</th>
<th>logD Ti</th>
<th>$C_3$ Ti</th>
<th>$C_4$ Ti</th>
<th>logD Al</th>
<th>$C_5$ Al</th>
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<td>18000</td>
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<td>-18.2 (0.13)</td>
<td>36.3 (5.4)</td>
</tr>
<tr>
<td>22_TiQn15</td>
<td>Solid</td>
<td>1400</td>
<td>4680</td>
<td>501 (16)</td>
<td>-18.14 (0.19)</td>
<td>0.37 (0.21)</td>
<td>85.4 (9.4)</td>
<td></td>
<td></td>
</tr>
<tr>
<td>23_TiQn16</td>
<td>Solid</td>
<td>1490</td>
<td>1500</td>
<td>655 (31)</td>
<td>-17.13 (0.13)</td>
<td>0.49 (1.13)</td>
<td>202.5 (33)</td>
<td>-16.73 (0.23)</td>
<td>54.5 (38.9)</td>
</tr>
</tbody>
</table>

Table S 1: Diffusion coefficients and background concentrations for Ti and Al in quartz, along with experimental temperatures and durations. The interface concentrations are extracted from a single fit to eq. (1), without incorporating the uncertainty on crater depth – these are provided as an approximation only, but are internally-consistent. The crater depths and uncertainties are taken from averages and standard deviations, respectively, of the central portion of the crater and the crystal surface adjacent to the crater, from concatenated east-west and north-south transects. The uncertainty on the diffusion coefficient incorporates the uncertainty on depth. Uncertainties are 1 s.d. Of the two missing Al data, one showed a highly irregular profile geometry so was not fitted, and the other showed no Al enrichment. Sol-gel source powders (referred to as Ti-doped SiO2) experiments are denoted TiQn##, solid-source powders are denoted TiQ## (referred to as TiO2-SiO2).

**INTERFACE CONCENTRATIONS**

Given our early experiences with roughened interfaces following experiments with pure rutile, all other experiments were done with low Ti powder sources. Ideally, the interface of the quartz crystal should have the same Ti (and Al) concentration as the source powder,
assuming that the loss of Ti from the powder as Ti diffuses into the crystal is negligible. This is not the case – we consistently see apparently low concentrations, except in two high temperature experiments where the interface concentrations are slightly higher than the concentrations in the powder. In the methods, we have attempted to account for this by assuming that the interface concentration could be either constant, or linearly change with time. If the interface concentration is not constant, then the linear change with time is also unlikely to be a reasonable assumption.

Herein, we test the potential effect of changing interface concentration on the extracted diffusion coefficients, in systems where there is a reasonable threshold below which concentration differences can no longer be distinguished. To do this, several numerical models were run (using an explicit finite difference approximation of Fick’s second law) in 1D, plane sheet geometry, with a constant diffusion coefficient \( \log_{10} D = -17.5 \text{ m}^2\text{s}^{-1} \) and time \( (10000 \text{ s}) \). For each model, a different relationship between the interface concentration and time was used. These included 1) no change as a function of time; 2) linear increase or decrease; 3) cubic increase or decrease and 4) sinusoidal oscillation. Several of these are unrealistic, but useful as a demonstration of concept. Models are shown in Fig. S 10.

In all models, the concentrations varied between 0 and 1, with 1 being the maximum possible interface concentration, and 0 always being the initial concentration. To extract an apparent diffusivity, the \( x=4\sqrt{D}t \) approximation of the effective diffusive length scale was used, which, for a perfect error function concentration decrease, with interface concentration of 1 and background concentration of 0, is the position at which the concentration \( (C(x)) \) decreases below 1-erf(1), i.e. \(~0.005\). Therefore, following each model, the x position closest to the point where \( C(x) = 0.005 \) was taken, and the apparent \( D \) was calculated using \( D=(x/4)^2/t \).
Fig. S 10 shows the models, as well as the extracted values of \( x \) where \( C(x)=0.005 \), and the associated \( \log_{10} D \) values. For the model where the interface concentration was constant, the extracted \( D \) (\( D^* \)) is the same as the input \( D \) (\( \log_{10} D=-17.5 \text{ m}^2\text{s}^{-1} \)). For most of the other interface conditions, there is an apparent decrease in the diffusivity. The largest differences between \( D^* \) and \( D \) are seen where there is an increase in the interface concentration as a function of time, with the largest difference observed where there is a cubic increase in interface concentration from 0 to 1. However, even in this very extreme case, the difference is 0.4 log units, far too small to explain the discrepancy between our data and the Cherniak et al. (2007) data. Moreover, if such an increasing interface concentration as a function of time were occurring, then we would expect to see a continuous increase in the Ti concentration as a function of experimental duration, which is not observed.
Fig. S 10: Modelling the effect of variable interface concentrations as a function of time on the extracted diffusion coefficients. The sub-plots in each row are related – the left plot shows the interface concentration as a function of time, the central plot shows the resulting diffusion profile (complete) and the right plot shows an enlargement of the central plot, considering only the portion where the concentration is between 0 and 0.025. The horizontal dashed line is drawn at concentration = 0.005, i.e. ~1-erf(2), and the vertical dashed line is drawn at the position where the horizontal line meets the modelled profile. The distance value (to the nearest nm) is shown, as is the associated diffusion coefficient calculated as $D=(x/4)^2/t$.

REFERENCES


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