

PECVD a-Si layers for industrial high-efficiency solar cell processing

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ABSTRACT

Amorphous silicon is one of the most effective materials in passivating silicon interfaces. At Fraunhofer ISE, highly passivating amorphous silicon coatings were developed by an industrially applicable Plasma-Enhanced Chemical Vapour Deposition (PECVD) process. Thin-film stacks of amorphous silicon and SiO_x display excellent passivation quality, indicated by effective charge carrier lifetimes ranging from 900 to 1600 μs and resulting surface recombination velocities between 9 and 3 cm/s⁻¹. The demonstrated temperature stability opens up new application opportunities also for amorphous silicon films in the industrial production of highly efficient solar cell structures, which will be further discussed in this paper.

Introduction

All cell structures that have shown efficiencies greater than 20% feature an efficient surface passivation with thin, mostly dielectric layers. However, the present state-of-the-art rear surface structure of industrial silicon solar cells is a screen-printed and thermally-fired Al back surface field (Al-BSF), which has two major restrictions: (i) the wafer bow due to mechanical stress, and (ii) the lower electrical and optical properties. A common measure for the quality of a surface is represented by the so-called 'surface recombination velocity' (SRV). The term 'surface recombination' describes the recombination processes taking place at the semiconductor surface. Here, the crystal lattice is abruptly disconnected, leading to a large amount of dangling bonds. These disturbances of the lattice create energy levels that are partly located within the band gap in between the conduction and the valence band (see Figure 1). The surface recombination velocity (SRV) S is defined as:

$$S = \frac{U_s}{\Delta n},$$

with U_s as surface recombination rate and Δn as excess carrier density. The larger the SRV, the quicker the charge carriers recombine at the surface.

When a dielectric layer is deposited on top of the surface, additional so-called 'band bending' typically occurs. The band bending is due to the difference in the Fermi level in the substrate and the layer as well as the fixed charges within that layer. The effective SRV S_{eff} accounts for the band bending effect by virtually shifting the surface into the wafer bulk to the location where the band bending starts, as outlined below:

$$S_{eff} = \frac{U_s}{\Delta n(x=d)}$$

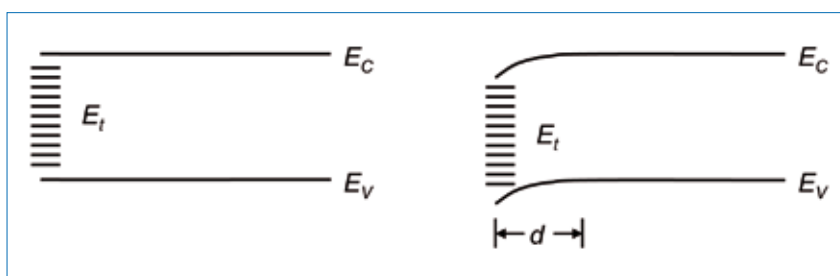


Figure 1. Defect (trap) energy levels within the band gap at the wafer surface (left picture: without band bending; right picture: with band bending).

There are two different mechanisms that lead to good surface passivation [1]: (i) the reduction of interface states D_{it} , and (ii) field effect passivation – that is, the strong reduction of one carrier type by incorporation of fixed charges Q_f in the passivation layer. Although these mechanisms or the combination of

both lead to low surface recombination velocities, the resulting $S_{eff}(\Delta n)$ curve shows different characteristics. The reduction of interface states is more effectively reached for thermally grown SiO₂ layers, while the field-effect passivation together with a moderate reduction of D_{it} is more

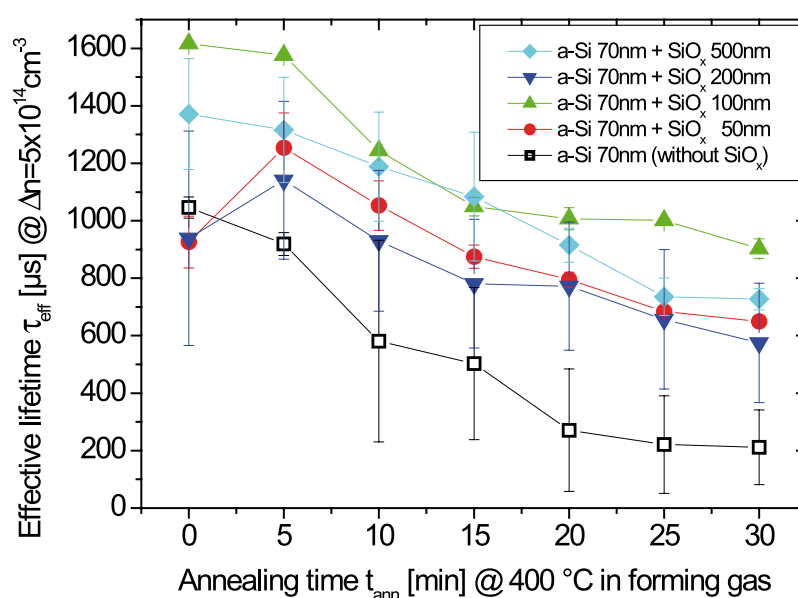


Figure 2. Carrier lifetimes of symmetric (SiO_x/a-Si/c-Si/a-Si (/SiO_x)) samples after stepwise annealing. a-Si + SiO_x systems show best stability [16].

typical for PECVD deposited layers like SiN_x . Typical values for SiO_2 are $\Delta_{it} = 10^{10} \text{cm}^{-2} \text{eV}^{-1}$ and $Q_f = 10^{10} \text{cm}^{-2}$, while for SiN_x , values are $D_{it} = 10^{11} \text{cm}^{-2} \text{eV}^{-1}$ and $Q_f = 10^{11} \text{cm}^{-2}$.

“Since the inversion layer is a crucial part of the surface passivation mechanism of SiN_x layers, the apparent quality of SiN_x layers on lifetime test wafers ‘vanishes’ when applied to real cells.”

Although it seems possible to design a perfect layer or layer system simply by performing lifetime and optical measurements, a final decision has to be made by applying these layers to solar cells. SiN_x passivation is an example of how good surface passivation quality is only a necessary but not a sufficient condition. Although single SiN_x layers show the very best surface passivation quality on lifetime test wafers – even better than thermal oxidation – none has so far managed to fabricate a cell with efficiencies attainable by those featuring the ‘classical’ thermal oxidation. In particular, the short-circuit current is significantly lower. This reduction was explained by Dauwe et al

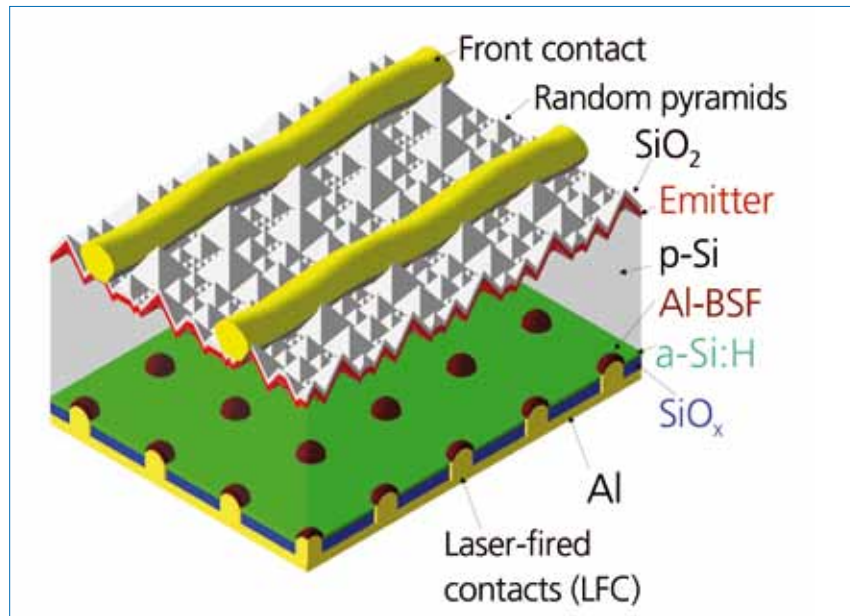


Figure 3. Schematic of fabricated solar cells. The rear surface passivation was created using a stack of PECVD a-Si:H and PECVD a-SiO_x:H.

[2] via the short-circuiting of the inversion layer induced by the fixed charges in the SiN_x layer at the rear contact points. Since the inversion layer is a crucial part of the surface passivation mechanism of SiN_x layers, the apparent quality of SiN_x layers on lifetime test wafers ‘vanishes’ when applied to real cells. This problem can be solved in two ways:

- the application of a dielectric stack system consisting of a thin silicon oxide

below the SiN_x layer. This oxide layer can even be deposited by PECVD (excellent surface passivation quality has been reported [3-6]).

- by the application of amorphous silicon as a passivation layer. For the HIT cell structure, it was a natural choice to use amorphous silicon as the rear surface passivation since the emitter is also formed by this layer type. This type of passivation is also applied on standard

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silicon solar cells achieving excellent cell results [7] reaching efficiencies up to 21.7% ($V_{oc} = 677\text{mV}$) [8].

The following section will focus on explaining the latter approach in particular in more detail.

Amorphous silicon

Lifetime investigation

Hydrogenated amorphous silicon (a-Si:H) layers have been in use for many years within the photovoltaic community. Thin-film solar cells deposited on glass substrates or deposited amorphous emitters on crystalline silicon wafers as found in the HIT (heterojunction with intrinsic thin layer) structure [9] can be found in industrial production. For passivation of crystalline silicon wafer surfaces, it has been shown that a Si:H layers can provide an extremely effective means to enhance the minority carrier lifetime [10-13]. However, one typical characteristic of a Si:H layers is their relatively low thermal stability, which limits the applicability of a Si:H passivation in industrial production lines. The incorporation of an additional PECVD a $\text{SiO}_x\text{:H}$ layer on top of the a Si:H leads to an improved thermal stability of the passivation quality of a Si:H at a c Si surface. Initially (as-deposited), all samples (Float Zone

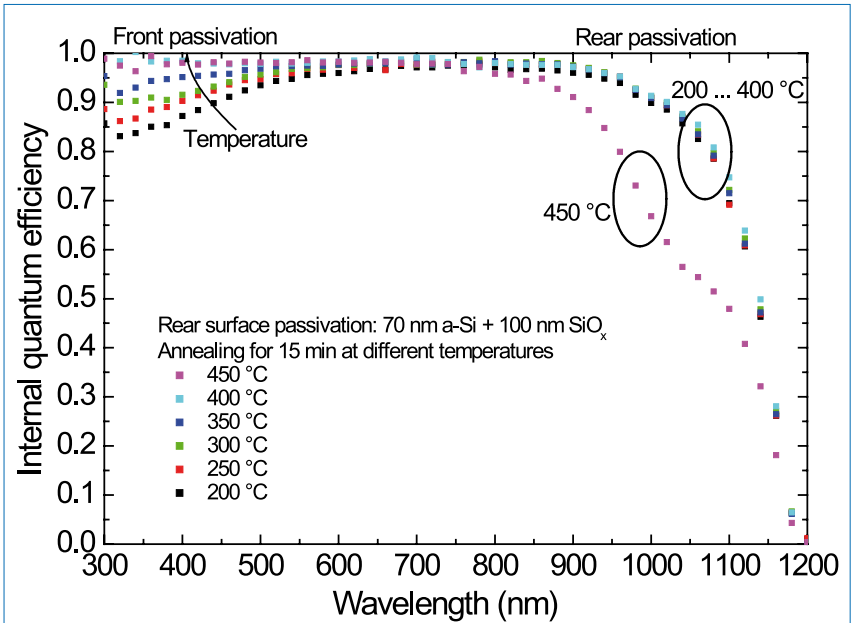


Figure 4. Internal quantum efficiency of solar cells annealed at different temperatures vs. the wavelength of light. An increase in front passivation (thermally-grown SiO_2) at short wavelengths with annealing temperature was found. The rear passivation (a-Si + SiO_x) is stable until 400 °C but decreases strongly at 450 °C [8].

silicon wafers of 1Ωcm, p-type (Boron-doped), 250μm-thick, shiny etched surfaces, both sides covered with a Si:H) showed an excellent surface passivation with lifetimes in the range of 900μs to 1600μs. As the maximum bulk lifetime

of these samples can be calculated to 2300μs when only Auger recombination is assumed (using the Auger model by Glunz and Rein [14]), the surfaces must be at a very well passivated state. Surface recombination velocities in the

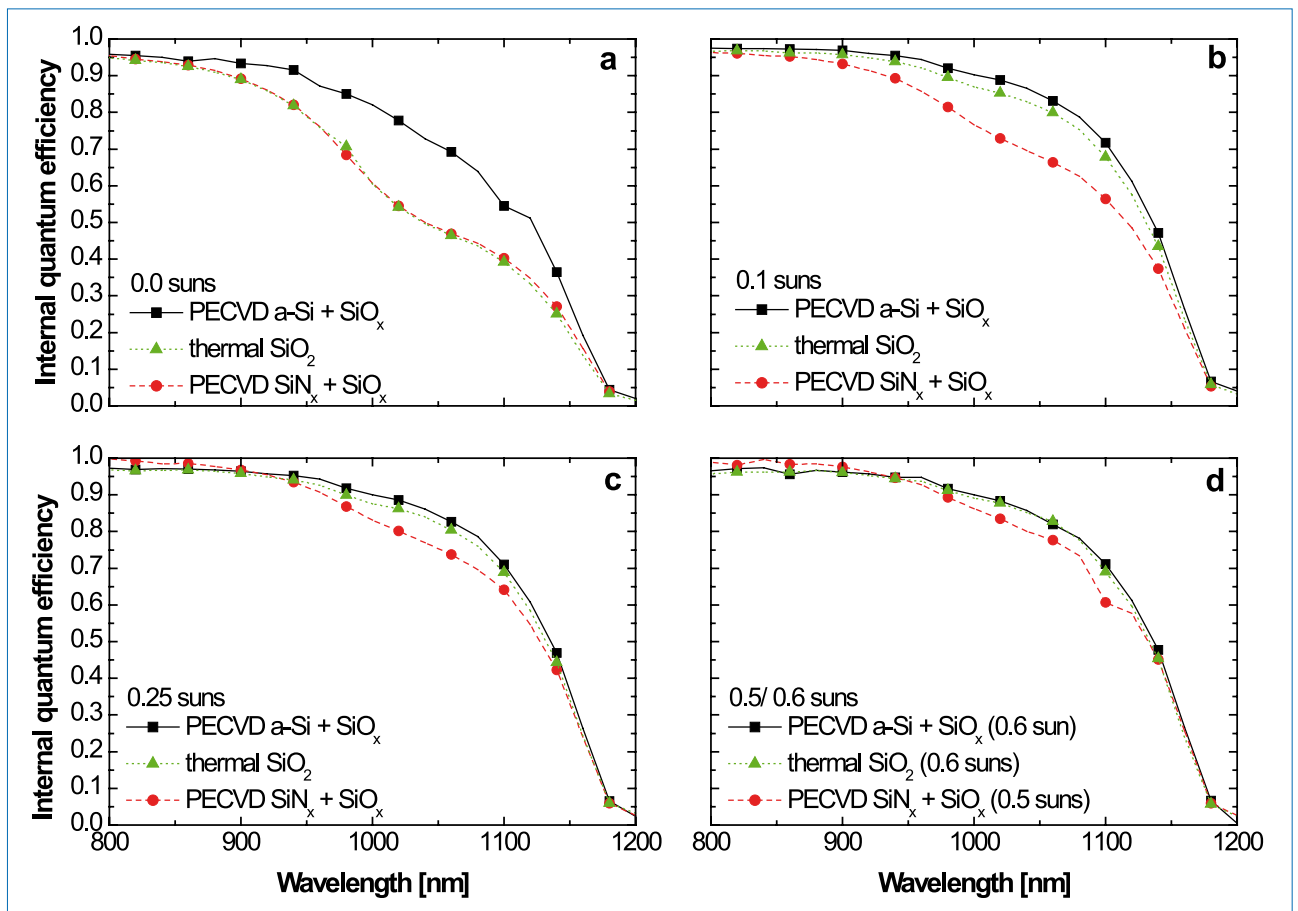


Figure 5. Comparison of bias light-dependent IQE measurements in the long wavelength regime of solar cells with (i) a-Si:H + a- $\text{SiO}_x\text{:H}$ passivation, (ii) thermally-grown SiO_2 passivation, and (iii) a- $\text{SiN}_x\text{:H}$ + a- $\text{SiO}_x\text{:H}$ passivation on the rear [16].

range of 9cm/s to 3cm/s can be calculated using the equation of Nagel et al [15]. No significant difference between the single-layer a-Si and the double-layer a-Si + SiO_x was observed. After annealing for 5 minutes in forming gas at 400°C, most lifetime results were quite stable. (One exception: the samples with lifetimes of approx. 900μs in the as-deposited state (samples with SiO_x thickness of 50nm and 200nm) improved significantly to between 1100μs and 1300μs. The samples with single-layer a-Si passivation degraded slightly to approx. 900μs.)

All lifetime values degraded with increasing annealing time. A stronger degradation could be found for the single layer a-Si passivated sample. After 30 minutes, an approximate lifetime of 200μs could be measured. The stack layer samples (a-Si + SiO_x) generally showed a better thermal stability compared to the single layer (a-Si) samples. No clear dependence could be found for the thickness of the SiO_x layer in the investigated thickness range.

Solar cell application

In order to show the capability of a-Si layers as a rear surface passivation layer, high-efficiency solar cells were fabricated. The cells exhibit evaporated TiPdAg front contacts, a thermally oxidised anti-reflection coating that also serves as the front passivation layer, a 120Ω/sq n-type emitter, a 0.5W/cm p-type bulk, PECVD-a-Si and -SiO_x, an evaporated Al layer at the back and laser-fired contacts that led to a local Al-BSF underneath the point contacts (see Figure 3).

“The stack layer samples (a-Si + SiO_x) generally showed a better thermal stability compared to the single layer (a-Si) samples.”

The a-Si (~70nm) and SiO_x (~100nm) layers used are the same as were investigated in the lifetime experiment. After finishing the cells, the I-V characteristics were measured. Subsequently, the cells were annealed at different temperatures in forming gas for 15 minutes (excluding 10 minutes of ramping up the temperature) to find the optimum annealing temperature for front and rear passivation and for the local laser-fired rear contacts. The challenge was to find a temperature that would increase the front passivation and the rear contacts without harming the rear passivation. The I-V-characteristics of the cells were then re-measured.

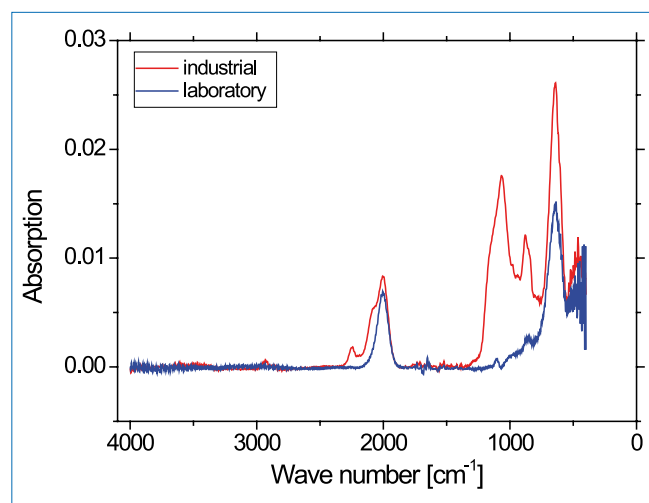


Figure 6. FTIR spectra of amorphous silicon layers deposited in a laboratory-type (blue) and an industrial-type PECVD (red) reactor [16].



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The solar cell parameters show a strong dependence on the annealing temperature; efficiency increases steadily for annealing temperatures from 200°C to 400°C with values of 20.0% to 21.7% with a strong decrease at temperatures above 450°C.

The question that arises here is: which physical effect drives the efficiency to these values? Internal quantum efficiency (IQE) measurements give insight into the answer to this question and are depicted in Figure 4. The cells were annealed at temperatures between 200°C and 450°C in steps of 50°C. A short wavelength regime was found in the range of 300nm to around 550nm affecting the solar cell's front surfac. The observed increase in quantum efficiency with higher annealing temperatures to about the same results at temperatures of 400°C and 450°C is attributed to the annealing of the front surface passivation, performed by thermal oxidation. Thus, the well-known oxide annealing effect has been described. The quality of the solar cell's bulk and rear surface can be derived from the IQE at long wavelengths around 900nm to 1200nm, where very high values can be found. Hence, the quality of the bulk and the rear surface seems very satisfactory.

For annealing temperatures of 200°C to 400°C the passivation properties of the rear surface is almost stable, with only the possibility of a slight improvement with higher applied temperatures. When the temperature is increased further to 450°C, the performance of the solar cell's back deteriorates compared to that recorded using lower temperatures. This effect fits in the observation of an overall decreased cell performance of the 450°C annealed cells at the I-V measurements. We attribute this effect to the de-passivation of the solar cell's back.

Hydrogenated amorphous silicon is known for its relatively low thermal stability. The observed de-passivation is most likely due to the cracking of hydrogen bonds at the c-Si a-Si interface, which are suspected to be important for the passivation effect. This leaves silicon dangling bonds and Shockley-Read-Hall (SRH) active recombination centres at the interface.

We found that the quantum efficiency behaviour of cells with a Si:H rear passivation did not change for bias light intensities of 0.1 to 0.6 suns. Therefore, a very stable rear passivation has been achieved that also provides excellent passivation at low light intensities. At 0.0 suns bias intensity the cell's rear showed slightly lower IQE values in the long wavelength regime. Hence, the rear passivation is significantly lower but still on a decent level. When the bias light dependence is compared to thermal SiO₂ or PECVD a SiN_x:H (very Si-rich), we found that the a-Si:H passivation is the most stable (see Figure 5).

Transfer to industrial deposition equipment

The deposition of high-quality passivating amorphous silicon in a small laboratory-type direct plasma PECVD reactor has been proved by Hofmann et al in 2004. For implementation of the a-Si passivation process, it is necessary to transfer the deposition process to high-throughput equipment. In our investigation, an SiNA machine from Roth&Rau was used, a PECVD tool that exhibits plasma excitation by microwaves that are introduced into the reactor by linear antennas.

“The minority carrier lifetimes were measured with the QSSPC technique leading to lifetimes up to >1ms ($S_{eff} < 10\text{cm/s}$).”

Usually, the topic that generates most interest in this regard is the surface passivation quality of a-Si layers deposited with the industrial PECVD tool. High-quality silicon wafers with the following characteristics were used: float zone, p-type, boron-doped, 1Ωcm, 250μm-thick, shiny etched surfaces. After cleaning the surfaces in a wet chemical RCA bath sequence, amorphous silicon was deposited sequentially on both surfaces. Next, the minority carrier lifetimes were measured with the QSSPC technique leading to lifetimes up to >1ms ($S_{eff} < 10\text{cm/s}$).

Fourier transformed infrared spectroscopy (FTIR) offers a means to characterise the composition of thin layers. In this case, a comparison between an a-Si layer deposited in our laboratory-type and our industrial-type PECVD reactor was performed, the results of which are shown in the graph in Figure 6.

The thickness of the layers is approximately the same (lab: 70nm, industrial: 85nm), which means that the peak heights can be directly compared. The absorption peaks at 2000cm⁻¹ are attributed to Si-H bonds. Hence, the Si-H bond density in both layers is approximately in the same order of magnitude.

In the wave number range of 1000cm⁻¹ the absorption peaks correspond to Si-O and Si-N bonds. It seems clear that the a-Si layer deposited in the industrial-type PECVD reactor is contaminated with nitrogen and oxygen in a stronger way than the reference a-Si layer. Nevertheless, this does not lower the surface passivation properties significantly.

Conclusion

Stacks of amorphous silicon and silicon oxide – both deposited applying a PECVD system – are successfully used for passivating crystalline silicon wafers, leading to surface recombination velocities below 6cm/s. These stacks were used to passivate crystalline silicon solar cells' rear surfaces and led to a maximum cell efficiency of 21.7% on p-type (Boron-doped) float zone silicon substrates with a thickness of 250μm.

I-V and IQE measurements lead us to the conclusion that the rear surface passivation (a-Si + SiO_x) was stable until a temperature of 400°C was reached. Bias light-sensitive IQE measurements showed very stable passivation properties of the a-Si:H/ a SiO_x:H stack at low injection levels. Excellent surface passivation could be reported for a-Si layers deposited in an industrial-type inline PECVD reactor. These a-Si layers exhibit an increased absorption in FTIR measurements, which can be attributed to N and O contamination.

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