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# Exploring the interplay of Ti-Sn co-doping in photoelectrochemical water splitting of hematite nanowires

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## Abstract

Photoelectrochemical water splitting is a promising alternative for sustainable energy production, addressing the growing need for clean energy sources. Hematite is a potential semiconductor for this process due to its abundance, low cost, non-toxicity, and stability. However, bare-hematite-based photoelectrochemical cells face challenges such as low photocurrent density, requiring innovative strategies to improve efficiency. This study explores the combined effects of three key approaches: enhancing crystallinity through high-temperature annealing, increasing specific surface area via nanostructuring, and improving photoanode conductivity through heteroatom doping. Hematite nanowires were synthesized using a hydrothermal method, with Ti-doping introduced during hydrothermal synthesis and subsequent Sn co-doping during an 800 °C annealing process, which also improved crystallinity. The introduction of Ti dopant significantly increased the photocurrent density under simulated solar illumination from 0.03 mA·cm<sup>-2</sup> to 0.63 mA·cm<sup>-2</sup>. Co-doping with Ti and Sn further enhanced performance to 1.27 mA·cm<sup>-2</sup>. The research explores how heteroatom doping influences the properties of hematite and examines its interaction with high-temperature annealing. These findings are significant for advancing the



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design of efficient nanostructures for energy conversion applications.

**Keywords:** Hematite, hydrogen production, heteroatom doping, nanostructuring, photoelectrochemistry, photoanodes

## INTRODUCTION

In recent years, rapid industrial growth and societal development have led to a significant increase in fossil fuel consumption<sup>[1]</sup>. Consequently, the world faces two major challenges: environmental pollution and energy resource depletion<sup>[2,3]</sup>. Direct production of hydrogen from renewable resources such as sunlight and water using a photoelectrochemical (PEC) cell holds promise as a cost-effective approach for harvesting and storing solar energy<sup>[4]</sup>. The use of hematite ( $\alpha\text{-Fe}_2\text{O}_3$ ) photoanodes in PEC water splitting has attracted significant interest because of its abundance, very good long-term stability<sup>[5]</sup>, and appropriate bandgap (about 2.2 eV), theoretically enabling a solar-to-hydrogen conversion efficiency of 16.8%<sup>[6-8]</sup>. However, pure-hematite-based PEC cells suffer from low photocurrent density and large overpotential, primarily due to hematite's poor conductivity and short hole diffusion length, resulting in high electron-hole recombination and efficiency losses<sup>[9]</sup>.

To improve the hematite photocurrent density, two parameters can be manipulated: recombination reaction time ( $\tau_r$ ) and carrier (hole or electron) drift time ( $\tau_d$ ). Increasing  $\tau_r$  (reducing recombination rate) can be achieved by passivating surface states to diminish the positive charge accumulation on the surface<sup>[10]</sup>. Additionally, improving crystallinity by reducing the number of grain boundaries, which act as non-radiative recombination centers, through thermal annealing processes is beneficial<sup>[11-13]</sup>. Conversely, decreasing  $\tau_d$  (accelerating carrier extraction) involves reducing carrier migration distance and enhancing carrier mobility to boost the built-in field.

Nanostructuring significantly enhances PEC performance by expediting photogenerated carrier extraction and improving charge separation<sup>[14,15]</sup>. Nanostructuring results in benefits such as larger surface area, better light absorption, reduced electron-hole recombination, and enhanced electronic properties due to quantum confinement effects<sup>[16-18]</sup>. Notably, the one-dimensional structure of nanowires (NWs), influenced by quantum confinement, shifts the absorption spectrum towards the visible range, thereby enhancing photoactivity.

The inclusion of impurity atoms, whether intentionally or unintentionally, within the hematite lattice has been demonstrated to significantly enhance photocurrent<sup>[12,19]</sup>. Incorporating high concentrations (ranging from 0.5% to 20% atomic) of impurities such as Si<sup>[20]</sup>, Sn<sup>[11]</sup>, Se<sup>[21]</sup>, Cr<sup>[22]</sup>, Pt<sup>[23]</sup>, Mn<sup>[24]</sup>, Cr<sup>[25]</sup>, Ge<sup>[26]</sup>, W<sup>[27]</sup> or Rh<sup>[28]</sup> has been achieved mostly through physical and chemical deposition, spray pyrolysis, or hydrothermal techniques<sup>[6,8,20,29-32]</sup>. Also, it has been shown that surface modification of hematite photoanodes can contribute to improving their PEC properties<sup>[33-36]</sup>. While most dopants yield modest improvements in PEC performance of hematite photoanodes, notable enhancements have been observed with Si, Ti and Sn doping<sup>[37]</sup>. In many studies, dopants are introduced into nanostructured hematite during the material growth process, allowing control over doping concentration. Particularly, Ti doping via hydrothermal techniques has become a primary method for improving hematite photocurrent by increasing donor density and carrier density, thus leading to higher solar-to-hydrogen conversion efficiency<sup>[38,39]</sup>.

While some studies have addressed Ti-doping<sup>[8,40-42]</sup>, most do not delve deeply into understanding the influence of Ti on the nanostructure. Furthermore, the high-temperature annealing strategy can

unintentionally dope the hematite photoanodes with Sn<sup>[11,12,43]</sup>. Although the effect of Sn doping has been explored, the interaction between Sn and Ti through intentional crossover doping has not been studied.

In this study, three strategies were combined to enhance hematite photoanode performance: (i) optimization of thermal annealing to enhance separation efficiency by increasing  $\tau_1$ , (ii) nanostructuring to decrease  $\tau_2$  by reducing carrier migration distance and improving specific surface area, and (iii) dual-doping to enhance carrier mobility, thereby increasing  $\tau_1$  and reducing  $\tau_2$ . For this purpose, hematite NWs obtained through the hydrothermal method were doped with Ti in a controlled manner during photoanode fabrication, while Sn-doping was achieved through Sn<sup>4+</sup> incorporation into the hematite structure from the fluorine-doped tin oxide (FTO) substrate during thermal annealing<sup>[12]</sup>.

This work aims to understand how the different strategies described above contribute to improving the efficiency of hematite photoanodes. For this purpose, this study successfully characterizes the impact of annealing conditions on dual doping with Ti and Sn and their interactions within the in-depth concentration profile using Rutherford Backscattering Spectroscopy (RBS). Moreover, the PEC performance evaluation of the photoanodes was conducted by determining the photocurrent density-voltage (*J-V*) curves and studying the incident photon-to-current efficiency (*IPCE*) as a function of wavelength. A comprehensive characterization of the photoanodes was also performed to better understand how each variable contributes to improving PEC cell efficiency. To this end, morphology was analyzed using field emission scanning electron microscopy (FE-SEM), composition was determined via X-ray photoelectron spectroscopy (XPS) and RBS, microstructure through X-ray diffraction (XRD) and Raman spectroscopy, and electrical properties using photoelectrochemical impedance spectroscopy (PEIS).

## EXPERIMENTAL

### Photoanodes synthesis

$\alpha$ -Fe<sub>2</sub>O<sub>3</sub> NWs, with and without Ti doping, were prepared by hydrothermal method<sup>[44]</sup>. FTO coated glass substrates (Solaronix, Aubonne), with dimensions 1.2 × 2.5 cm<sup>2</sup>, were previously cleaned as reported by Francisco *et al.*<sup>[45]</sup>. 7 Ω-square<sup>-1</sup> and 10 Ω-square<sup>-1</sup> substrates were used for the samples that followed the one-step and two-step annealing process, respectively, since the 7 Ω-square<sup>-1</sup> substrates were deformed when the high temperature annealing step was applied. Two schemes of samples were prepared by hydrothermal process (i) to obtain bare hematite NWs and (ii) Ti-doped hematite NWs. (i) FTO substrates were immersed in a stainless steel autoclave with 20 mL solution of 0.15 M FeCl<sub>3</sub>·6H<sub>2</sub>O and 1 M NaNO<sub>3</sub>, adjusted at pH 1.5 with HCl 37%<sup>[11,12]</sup>, and heated at 95 °C for 4 h<sup>[15]</sup>. (ii) In the case of Ti-doped samples, FTO substrates were also immersed in the solution (i), after adding 10 μL of Ti isopropoxide. Afterward, the samples were heated at 95 °C for 13 h (based on the procedure described by Yang *et al.*<sup>[12,46]</sup>). All reagents were bought from Sigma-Aldrich.

Two annealing treatment approaches were studied: the one-step annealing approach, which involves annealing at 600 °C for 3 h, and the two-step annealing approach, which includes a first step at 550 °C for 2 h, followed by a short annealing step at 800 °C for 20 min<sup>[12]</sup>. As-prepared samples with and without Ti dopant were subjected to both the one- and two-step annealing approaches. For Ti-doped samples, after the thermal annealing, the samples were again placed in an autoclave with solution (i) and heated at 95 °C for 20 min<sup>[46]</sup>. The undoped one- and two-step annealed samples are denoted as 600PA and 800PA, respectively, while the doped samples are denoted as Ti600PA and Ti800PA, respectively.

## Photoanodes characterization

### *Morphological characterization*

The morphological characterization was performed using FE-SEM (FEI Quanta 400 ESEM FEG, FEI, Oregon) at CEMUP, University of Porto. The diameter ( $D$ ) and thickness of NWs were estimated using ImageJ software<sup>[47]</sup>.

### *Structural characterization*

The structural analysis was performed using XRD (Rigaku SmartLab, Rigaku, Tokyo) with Cu K $\alpha$  radiation (1.540593 Å) in parallel-beam mode with grazing incidence omega angles. Measurements were made in a  $2\theta$  range from 20° to 80°, with 0.02° increments and 0.5 s accumulation time. Raman spectroscopy was also used to confirm the crystalline phase and ordering of the photoanodes. The Raman scattering was measured using a Renishaw inVia Qontor Spectrometer, using an excitation laser with a 633 nm polarized line of He-Ne, in the spectral range from 200 cm<sup>-1</sup> to 1,400 cm<sup>-1</sup>, an exposure time of 60 s and power of 1.1 mW. The diffraction patterns were refined by applying a Rietveld Refinement process, using the Match software (Crystal Impact, Bonn).

### *Surface compositional characterization.*

To study the surface composition of the samples, XPS characterization was carried out. A Phoibos 150 EP MCD analyzer was used (SPECs Group, Berlin). The undoped 800 °C sample was irradiated with Al K $\alpha$  X-rays ( $h\nu = 1486.6$  eV), while for Ti-doped samples, a Mg K $\alpha$  X-rays ( $h\nu = 1253.6$  eV) source was used. According to the Seah and Dench model for inorganic compounds<sup>[48]</sup>, sampling depths of 9.6 nm and 9.0 nm have been estimated when using Al and Mg K $\alpha$  X-rays, respectively. The take-off angle was fixed at 90° to the sample surface. Pass energies of 50 eV and 20 eV were used for the survey and core level spectra, respectively. The binding energy (BE) values were corrected for charging effects by referencing the adventitious carbon C 1s peak, at 284.8 eV.

### *Bulk compositional characterization*

On the other hand, to obtain additional information about the bulk composition, thickness and roughness, Rutherford ion backscattering experiments were performed with a 5 MV terminal voltage Tandetron accelerator located at “Centro de Microanálisis de Materiales”<sup>[49]</sup>. Experiments were conducted using a 4.260 MeV He<sup>+</sup> beam. The experiment was performed at normal incidence; backscattered ions were detected at 170°, and a cumulative charge of 12  $\mu$ C was collected. The experimental spectra were fitted using the SIMNRA 7.03 software<sup>[50]</sup>.

### *Photoelectrochemical impedance spectroscopy*

The electrical properties of the photoanodes were examined by PEIS. In this technique, a small sinusoidal potential perturbation is applied to the system and the amplitude and phase shift of the resulting current response are recorded. A BioLogic SP-150 potentiostat (Bio-Logic, Seyssinet-Pariset, France) was employed. A frequency range between 0.1 Hz and 100 kHz was covered with a modulation signal of 10 mV<sup>[55]</sup>. The EIS spectra were then fitted to an appropriate electrical equivalent circuit using ZView software (Scribner Associates Inc., Southern Pines).

### *Photoelectrochemical performance*

The PEC performance of the hematite photoanodes was assessed based on the  $J$ - $V$  curves, stability tests, intrinsic solar-to-chemical (ISTC) conversion efficiency, and IPCE measurements.



The photocurrent  $J$ - $V$  curves were performed using a class B solar simulator with a 150 W Xe lamp (Oriel, Newport) under dark and 1-sun simulated sunlight ( $100 \text{ mW}\cdot\text{cm}^{-2}$ , AM 1.5 G). The photoanodes were immersed in 1.0 M NaOH electrolyte solution using a “cappuccino” PEC cell<sup>[5,51]</sup>, with an illuminated surface area of  $0.528 \text{ cm}^2$  and a three-electrode configuration: Ag/AgCl/saturated KCl electrode as the reference electrode, a platinum wire as the counter-electrode and the hematite NW photoanodes as the working electrode. The photocurrent was assessed under front-side illumination, which resulted in higher photocurrent than backside illumination, as the transport of generated electrons from the illuminated hematite through the FTO layer is easier than transport of holes (with a short diffusion length of 2-4 nm) through the hematite layer<sup>[12,52,53]</sup>.

The stability tests were performed at a constant voltage of  $1.23 V_{\text{RHE}}$  (voltage vs. reversible hydrogen electrode) for two hours, using the same conditions described above. The *ISTC* conversion efficiency was determined from the photocurrent density vs. potential curves measured under dark and illumination, in a three-electrode configuration<sup>[5,54]</sup>. According to Rothschild’s model<sup>[54]</sup>, the *ISTC* is defined by

$$\begin{aligned} \text{ISTC} &= \left[ \eta_{\text{el}} \frac{J_{\text{photo}}(\text{mA}\cdot\text{cm}^{-2}) \times V_{\text{photo}}(\text{V})}{P_{\text{light}}(\text{mW}\cdot\text{cm}^{-2})} \right]_{\text{AM1.5G}} \\ &\cong \frac{1.23(V_{\text{RHE}})}{U_{\text{dark}}(V_{\text{RHE}})} \left[ \frac{J_{\text{photo}}(\text{mA}\cdot\text{cm}^{-2}) \times V_{\text{photo}}(\text{V})}{100(\text{mW}\cdot\text{cm}^{-2})} \right]_{\text{AM1.5G}} \end{aligned} \quad (1)$$

with  $\eta_{\text{el}}$  the electrolysis efficiency,  $J_{\text{photo}}$  the generated photocurrent,  $V_{\text{photo}}$  the correspondent photopotential and  $U_{\text{dark}}$  the potential applied to the photoanode to reach the respective dark current<sup>[5,54]</sup>. Furthermore, the intrinsic photovoltaic power  $P_{\text{light}}$  for the photoanodes is given by

$$P_{\text{light}}(\text{mW}\cdot\text{cm}^{-2}) = J_{\text{photo}}(\text{mA}\cdot\text{cm}^{-2}) \times V_{\text{photo}}(\text{V}) \quad (2)$$

Also, the fill factor (*FF*) at the maximum power point (*MPP*) is determined by

$$\text{FF} = \frac{V_{\text{MPP}} \times J_{\text{MPP}}}{V_{\text{OC}} \times J_{\text{SC}}} \quad (3)$$

where  $V_{\text{MPP}}$  and  $J_{\text{MPP}}$  indicate the values of photopotential and photocurrent at the *MPP*, respectively;  $V_{\text{OC}}$  and  $J_{\text{SC}}$  represent the open circuit voltage and short circuit current, respectively.

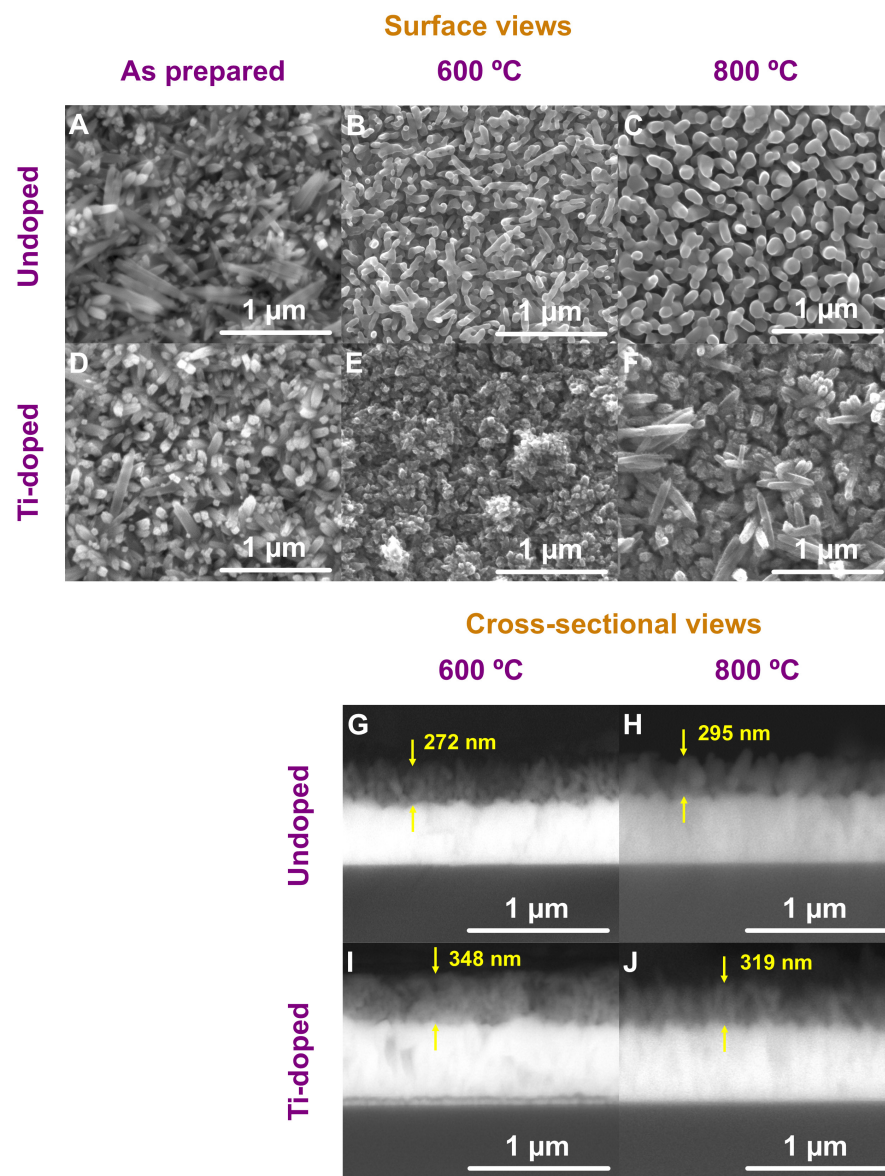
*IPCE* measurements were carried out using a 150 W continuous-wave Xe lamp, with a computer-controlled monochromator. A scan was performed for wavelengths between 330 nm and 650 nm. The scanning speed was  $2 \text{ nm/s}$ . Measurements were performed at  $V_{\text{RHE}} = 1.45 \text{ V}$ . The *IPCE* can be calculated from the photocurrent measured at each wavelength, based on the expressions proposed in previous reports<sup>[4,7]</sup>.

## RESULTS AND DISCUSSION

### Morphological characterization

The morphology of the prepared photoanodes was assessed through FE-SEM, as shown in Figure 1A-F. The as-prepared NWs were observed to be fully formed during the initial stage of the autoclaving process. This formation is attributed to the thermodynamically stable nucleation of  $\text{Fe}^{3+}$  ions in solution, under controlled temperature and pressure conditions<sup>[44]</sup>.

For the undoped samples, following a one-step annealing process at  $600^\circ\text{C}$ , the NWs exhibited a rounder morphology. Furthermore, the  $D$  of the NWs notably increased when subjected to a second annealing step



**Figure 1.** FE-SEM surface images of (A) undoped NWs after hydrothermal synthesis; and after (B) one-step and (C) two-step annealing processes; (D) Ti-doped NWs after hydrothermal synthesis, and after (E) one-step and (F) two-step annealing processes; Cross-sectional views of undoped (G) one-step and (H) two-step annealed, as well as Ti-doped (I) one-step and (J) two-step annealed  $\alpha$ -Fe<sub>2</sub>O<sub>3</sub> NWs. FE-SEM: Field emission scanning electron microscopy; NWs: nanowires.

at 800 °C for 20 min. Specifically, *D* measurements from top-view FE-SEM images revealed an increase from  $63 \pm 9$  nm to  $97 \pm 19$  nm when considering the one-step and the two-step annealing processes, respectively. The observed increase is attributed to the fusion between neighboring NWs during the high-temperature annealing step<sup>[11,12,56]</sup>.

In the case of Ti-doped photoanodes, a less ordered structure was obtained after one-step annealing at 600 °C compared to the undoped samples. In fact, the morphology and ordering of the NWs reflects a more random arrangement of hematite nanoparticles. Similar effects have been reported in previous works<sup>[57,58]</sup>. Moreover, the trend observed in undoped samples regarding NW diameter dimensions persisted for the

Ti-doped samples in both annealing schemes:  $D$  increased from  $46 \pm 15$  nm to  $79 \pm 8$  nm when employing the two-step annealing process.

Figure 1G-J shows FE-SEM cross-sectional views of the samples. The thickness of the hematite photoanodes was estimated, revealing that for the undoped samples, the hematite layer measures approximately 272 nm after annealing at 600 °C. The thickness slightly increases to 295 nm following the final annealing step. In contrast, Ti-doped samples exhibit a higher thickness of about 348 nm when annealed at 600 °C. However, the high-temperature annealing step at 800 °C reduces the thickness in this case to 319 nm. In any case, the determination of the thickness of the hematite layer for such inhomogeneous samples is subject to a considerable margin of uncertainty. It can also be observed that in the case of the Ti-doped samples, the structure is more spongy and less ordered than in the case of the undoped ones, consistent with what was observed from the surface views.

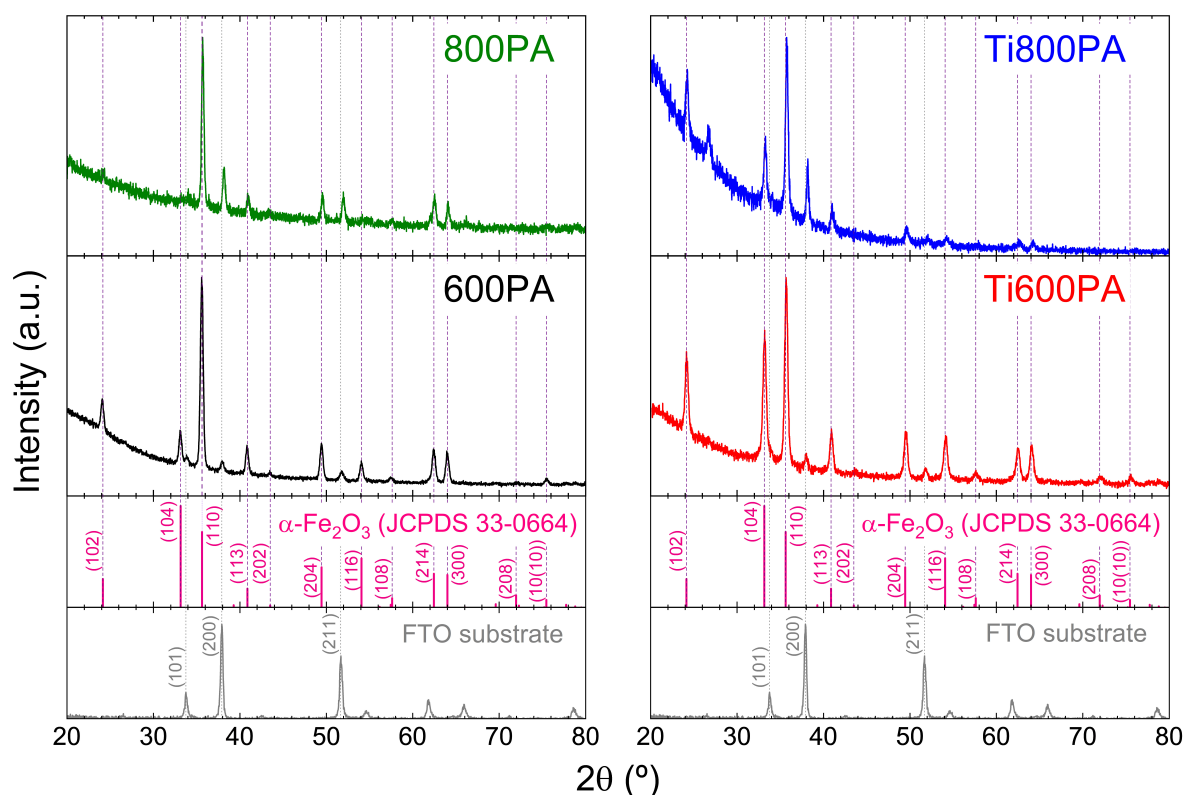
### Structural analysis

Figure 2 shows the XRD patterns for both bare and Ti-doped hematite NWs, subjected to the one- and two-step annealing schemes. It is observed that, for all the samples, the diffraction pattern contains peaks associated with the crystalline phase of hematite ( $\alpha$ -Fe<sub>2</sub>O<sub>3</sub>), as well as with the main peaks of the FTO substrate. In all cases, the (110) diffraction peak predominates, suggesting that the crystallites are highly oriented vertically with respect to the substrate. Supplementary Table 1 shows the ratio between the intensity of the (110) and (104) diffraction peaks. This ratio is higher for the 800 °C annealed samples than for the 600 °C annealed samples, as well as for the undoped samples compared to the doped ones. This is consistent with what has been reported in the literature and confirms that the incorporation of Ti during the hydrothermal synthesis reduces the selectivity of the nucleation, leading to a more random distribution, in agreement with SEM images<sup>[58-60]</sup>.

The  $\alpha$ -Fe<sub>2</sub>O<sub>3</sub> lattice structure consists of an alternation of iron bilayers and oxygen layers, aligned parallel to the (001) basal plane. Within each bilayer, Fe<sup>3+</sup> atoms have parallel spins, while the spins in adjacent bilayers are antiparallel<sup>[61,62]</sup>. Electrons can move by hopping within the iron bilayers, but electron exchange between neighboring bilayers is spin forbidden by Hund's rule. Consequently, the conductivity of hematite is found to be four orders of magnitude higher within the (001) basal plane (e.g., in [110] direction) than in the orthogonal directions<sup>[61,62]</sup>. Therefore, having a preferential orientation of the crystallites along the [110] direction is expected to enhance the efficiency of the PEC process<sup>[63]</sup>.

In addition to the peak associated with plane (110), the peaks associated with planes (102), (104), (113), (204), (116), (108), (214) and (300) can be also appreciated.

The crystallite size  $D$  and inhomogeneous strain  $\epsilon$  have been estimated for each sample using the Williamson-Hall model<sup>[64,65]</sup>, as explained in the supplementary information. Williamson-Hall plots are shown in Supplementary Figure 1. With the two-step annealing scheme, an increase from  $D = 27 \pm 4$  nm to  $D = 37 \pm 12$  nm for the undoped and from  $D = 23 \pm 3$  nm to  $D = 30 \pm 5$  nm for the Ti-doped photoanodes is observed. This is ascribed to the hematite crystallinity improvement promoted by temperature<sup>[11,13,29]</sup>. For Ti-doped samples, smaller crystallites were obtained, as previously reported by other authors<sup>[59,66]</sup>. Also, the inhomogeneous strain has been estimated in  $\epsilon = 0.06 \pm 0.05$  %,  $\epsilon = 0.11 \pm 0.08$  %,  $\epsilon = 0.06 \pm 0.04$  %, and  $\epsilon = 0.12 \pm 0.05$  % for the 600PA, 800PA, Ti600PA, and Ti800PA photoanodes. Therefore, the 800 °C annealed samples present a slightly higher strain than the 600 °C samples. The greater strain for the 800 °C annealed samples than for the 600 °C annealed samples can be due to a greater incorporation of Sn<sup>4+</sup> in the  $\alpha$ -Fe<sub>2</sub>O<sub>3</sub> structure at the interface with the FTO when the annealing temperature is higher. However, since



**Figure 2.** XRD patterns of the undoped and Ti-doped hematite NWs, subjected to one-step (600 °C) and two-step (800 °C) annealing processes. The  $\alpha$ -Fe<sub>2</sub>O<sub>3</sub> reference pattern (JCPDS 33-0664), and the diffractogram corresponding to the FTO substrate are also shown. XRD: X-ray diffraction; NWs: nanowires; FTO: fluorine-doped tin oxide.

$\epsilon < 0.2\%$  for all samples, strain is not expected to have a relevant effect on the physical properties of the photoanodes<sup>[67–69]</sup>.

The cell parameters obtained for the hematite structure by the Rietveld Refinement, and the calculated unit cell volume  $V$  are shown in Table 1. It is observed that the undoped sample subjected to a two-step annealing process presents a slightly larger unit cell than the sample subjected to a one-step annealing process. However, this difference is minimal and could be attributed to the incorporation of Sn<sup>4+</sup>, which has an ionic radius of 0.69 Å, slightly larger than that of Fe<sup>3+</sup>, at 0.65 Å, within the structure<sup>[70]</sup>. In the case of the Ti-doped samples, while the unit cell volume for the sample subjected to a single-step annealing process is comparable to that of the undoped sample, a significantly greater reduction in unit cell volume is observed when the annealing occurs at 800 °C. This is likely due to the incorporation of Ti<sup>4+</sup>, which has an ionic radius of 0.61 Å, smaller than that of Fe<sup>3+</sup>, and suggests that the higher temperature of 800 °C facilitates a more effective integration of Ti into the  $\alpha$ -Fe<sub>2</sub>O<sub>3</sub> structure compared to the 600 °C annealing process<sup>[70]</sup>.

Supplementary Figure 2 shows the Raman spectra of different synthesized photoanodes. The spectra are consistent with the  $\alpha$ -Fe<sub>2</sub>O<sub>3</sub> structure, with the peaks at 244 cm<sup>-1</sup>, 293 cm<sup>-1</sup>, 410 cm<sup>-1</sup> and 612 cm<sup>-1</sup> corresponding to doubly degenerate E<sub>g</sub> modes and those at 225 cm<sup>-1</sup> and 497 cm<sup>-1</sup> to single degenerate A<sub>1g</sub> modes<sup>[71–73]</sup>. The full width at half maximum (FWHM) of each peak is shown in Supplementary Table 2. A greater broadening of the peaks is observed for the Ti-doped samples. This is consistent with the smaller crystallite size observed for the Ti-doped samples in relation to the undoped ones.

**Table 1. Cell parameters obtained from the Rietveld Refinement for each of the samples. The unit cell volume is also shown**

| Sample                   | 600PA   | Ti600PA | 800PA   | Ti800PA |
|--------------------------|---------|---------|---------|---------|
| <i>a</i> (Å)             | 5.0299  | 5.0295  | 5.0322  | 5.0174  |
| <i>c</i> (Å)             | 13.7335 | 13.7467 | 13.7339 | 13.7085 |
| Volume (Å <sup>3</sup> ) | 300.9   | 301.1   | 301.2   | 298.9   |

### Surface composition

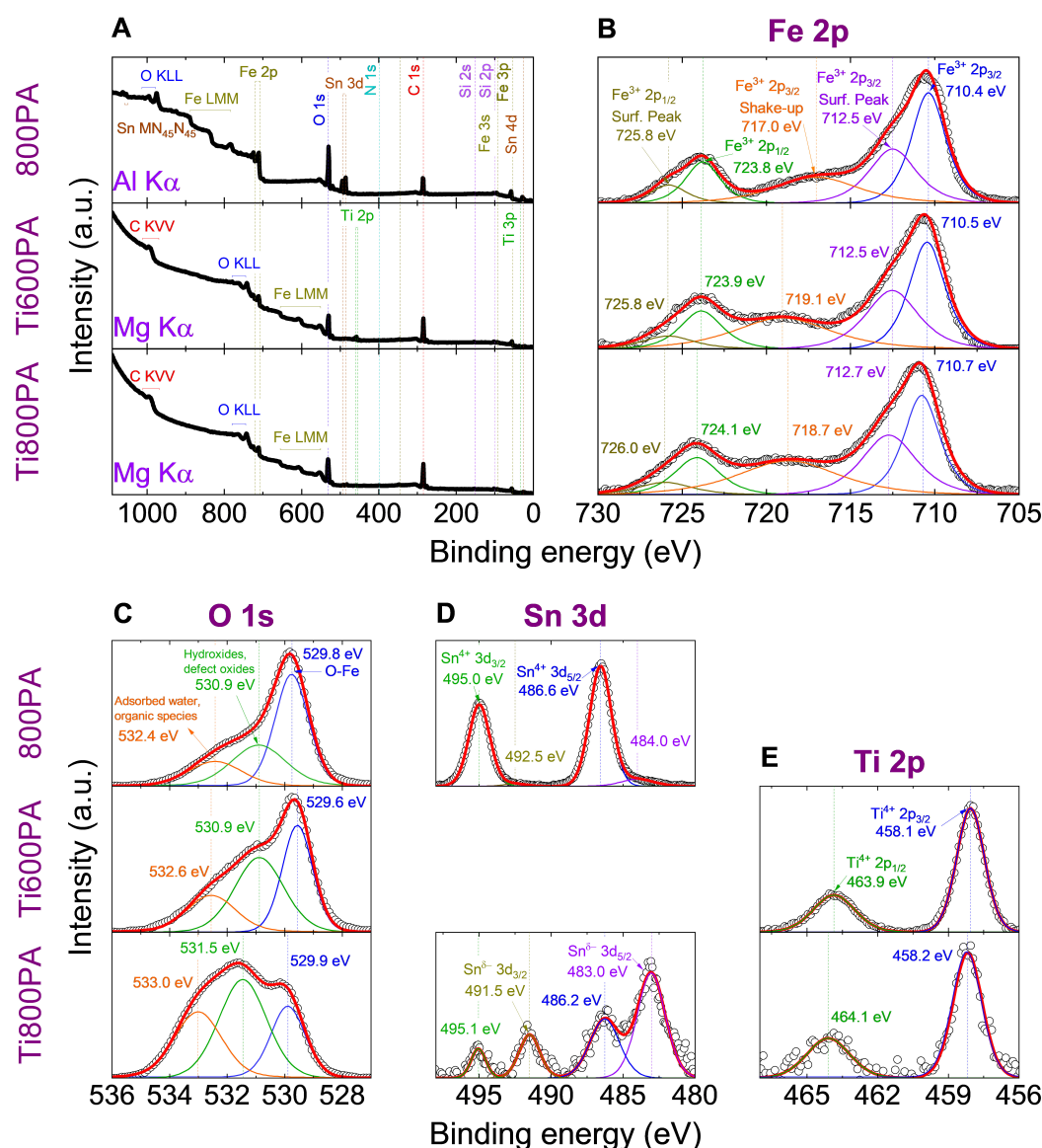
To assess the surface composition, an XPS study was performed on the 800PA, Ti600PA and Ti800PA photoanodes. The full survey for each sample and the high-resolution spectra corresponding to Fe 2p, O 1s, Sn 3d and Ti 2p are shown in Figure 3. For high-resolution spectra, the background was considered by applying a Shirley baseline model<sup>[74]</sup>.

In the full survey spectra, shown in Figure 3A, it is observed that the most intense signals correspond to the characteristic Fe and O peaks of hematite. Also, presence of Sn coming from the diffusion from the FTO substrate can be observed, along with Ti, for the samples doped with this element. In addition, a significant amount of C is observed, mainly due to adventitious carbon contamination during air exposure. In addition, traces of Si are observed at a very low concentration, probably due to small impurities introduced by the glass substrate itself. Also, a very small signal associated with N is appreciated, probably due to traces of the precursors.

The spectral region associated with the Fe 2p photoemission peak is shown in Figure 3B. Two split spin-orbit components are observed, located at 710.4-710.7 eV (Fe 2p<sub>3/2</sub>) and at 723.8-724.1 eV (Fe 2p<sub>1/2</sub>), which are consistent with the reported BE in  $\alpha$ -Fe<sub>2</sub>O<sub>3</sub><sup>[75-77]</sup>. Furthermore, the separation in energy of the two spin orbital components is 13.4-13.5 eV, which clearly verifies the formation of Fe<sub>2</sub>O<sub>3</sub><sup>[71]</sup>. No significant shift was observed between the different samples, and the small difference between the position of the different peaks could be due to small calibration errors of the analyzer, since it has been reported that using adventitious carbon for charge referencing can induce an error of the order of 0.25 eV in the position of the rest of the peaks<sup>[78]</sup>. In addition, surface components, associated with a higher BE, have been observed for both Fe 2p<sub>1/2</sub> and Fe 2p<sub>3/2</sub> photoemission peaks. These structures present BE of ~713 eV and ~726 eV, respectively. These surface peaks may indicate an atomic disorder on the surface of the material, which leads to a low electron density around the Fe(III) atoms. Consequently, the energy required to produce photoelectrons from these atoms is greater<sup>[79]</sup>. The atomic percentage of Fe corresponding to the surface component is approximately 38%, 41% and 43% for the 800PA, Ti600PA and Ti800PA samples, respectively. Therefore, no significant dependence is observed on the annealing temperature, or on the doping with Ti. Also, an additional very broad peak is observed at an energy of 717-719 eV, corresponding with the shake - up satellite associated with the Fe 2p<sub>3/2</sub> photoemission peak from  $\alpha$ -Fe<sub>2</sub>O<sub>3</sub><sup>[80,81]</sup>.

The O 1s spectral region is shown in Figure 3C. This region has been fitted with three Gaussian - Lorentzian peaks corresponding to lattice oxygen at the lowest BE (mainly Fe-O), to defective surface oxides or surface hydroxides at the medium BE, and to absorbed water or organic O at the highest BE<sup>[82,83]</sup>. The Fe-O peak is located at 529.6-529.9 eV, which is consistent with the reported BE in  $\alpha$ -Fe<sub>2</sub>O<sub>3</sub><sup>[39,84]</sup>. It is also worth mentioning that the percentage of the oxygen that forms the Fe-O bond, and the total oxygen is 51 %, 37 % and 23 % for the 800PA, Ti600PA and Ti800PA samples, respectively. Since the XPS measurements were performed after the PEC study, these differences may simply be due to small variations in the amount of -OH groups that have remained adhered to the surface, or to water or organic solvent molecules that may have been absorbed in the sample after the cleaning process.





**Figure 3.** XPS spectra collected from the 800PA, Ti600PA and Ti800PA photoanodes: (A) full survey; (B) Fe 2p; (C) O 1s; (D) Sn 3d; (E) Ti 2p. For high-resolution spectra, the background was deduced by applying a Shirley baseline model. XPS: X-ray photoelectron spectroscopy.

In Figure 3D, the spectral region corresponding to Sn 3d is shown. For the 800PA, two peaks at ~486.6 eV and ~495.0 eV are observed, corresponding to the photoemission peaks associated with the 3d<sub>5/2</sub> and 3d<sub>3/2</sub> levels of Sn, respectively, which are consistent with the Sn<sup>4+</sup> oxidation state<sup>[85,86]</sup>. Additionally, two weak peaks are observed at lower BE than the previous ones. The BE of these peaks is significantly lower than that expected for metallic tin (485.0 eV for the 3d<sub>5/2</sub> peak)<sup>[87]</sup>, so they must be negatively charged Sn species. Similar effects have been reported by other authors<sup>[88,89]</sup>. For the 800PA sample, more than 93% of the Sn is in the Sn<sup>4+</sup> state; in the case of the Ti800PA sample, this percentage is reduced to 36%, with the remaining 64% corresponding to negatively charged Sn species. In both cases, no metallic Sn is observed. The origin of these negatively charged species may be due to the incorporation of some Sn ions into the hematite structure, replacing O, or to the formation of some SnNa<sub>x</sub> species on the surface during PEC measurements. In this case, Na may not be observed in the survey spectrum because its sensitivity factor is four times lower



than that of Sn, whose concentration is already low. For the Ti600PA sample, the amount of Sn on the surface is lower than the detection limit of the technique, suggesting a barrier effect of the Ti doping, which prevents Sn diffusion towards the photoanode surface.

Finally, [Figure 3E](#) shows the spectral region corresponding to Ti 2p, for the Ti-doped samples. In both cases, two peaks at  $\sim 458.1$  eV and  $\sim 464.0$  eV are observed, corresponding to the photoemission peaks associated with Ti levels  $2p_{3/2}$  and  $2p_{1/2}$ , respectively. For titanium (IV) oxide, the position of the  $2p_{3/2}$  peak would be expected to be located at  $\sim 458.7$  eV. This shift to lower BE has also been reported by other authors for Ti-doped hematite<sup>[41,59,90-93]</sup>. This suggests that the Ti atoms are in a different chemical environment compared to the  $\text{TiO}_2$  structure, confirming their incorporation into the  $\alpha\text{-Fe}_2\text{O}_3$  crystal structure<sup>[59,93]</sup>. Also, in both cases, there is a separation of 5.8-5.9 eV between the  $2p_{3/2}$  and  $2p_{1/2}$  peaks, which confirms that the titanium is in  $\text{Ti}^{4+}$  state.

A quantitative estimation of the surface composition has been performed based on the intensity of the signals associated with each element. For this, the sensitivity factors provided by the manufacturer of the analyzer have been considered. When conducting this calculation, the carbon and oxygen that do not form Fe-O bonds have not been considered, since they are essentially surface features present on the outfacing side of the sample. Thus, [Table 2](#) shows the relative concentration of  $\text{O}_{\text{lattice}}$ , Fe, Sn and Ti for each of the samples.

A higher concentration of Ti on the surface is observed for an annealing temperature of 600 °C (9.2 at.%) compared to 800 °C (1.4 at.%). Furthermore, it should be noted that for the two samples annealed at 800 °C, the Sn concentration in the undoped sample is about seven times higher than in the Ti-doped sample. This variation may be attributed to the more defined morphology and ordering of the NWs of the undoped sample. As a result, part of the analyzed area may correspond to an area with a very low hematite thickness, facilitating better diffusion of Sn towards the surface for this sample (no Ti barrier).

#### **In-depth compositional characterization.**

While XPS is a very useful technique to determine with great precision the composition and chemical state of the different elements present on the outmost (5-10 nm) of the sample, using Rutherford Backscattering Spectrometry (RBS) makes it possible to determine the in-depth distribution of the different elements present in the photoanodes<sup>[94,95]</sup>.

[Supplementary Figure 3](#) shows the RBS spectra obtained for the 800PA, Ti600PA and Ti800PA samples. Only the relevant part of the spectra, corresponding to the hematite photoanode and the FTO layer, is shown. To simulate the diffusion of Sn from the FTO substrate and the incorporation of Ti, the experimental data have been fitted, simulating up to twenty layers corresponding to hematite doped with different atomic percentages of Ti and Sn, on top of up to thirteen more layers containing Sn and O to simulate the FTO substrate. C was also considered when performing the simulations to consider possible organic contamination present in the samples.

According to the highest energy edge of the Fe signal, located at  $\sim 3,205$  keV for all the samples, the presence of this element on the surface is identified. Furthermore, for the Ti-doped samples, the presence of Ti on the surface is also observed in both samples, as per its high energy edge, located at  $\sim 3,056$  keV. The lower energy edges of the Fe and Ti signal, along with the high energy edge of the Sn signal, are quite smooth. This effect is particularly prominent in the undoped sample annealed at 800 °C. This phenomenon results from a combination of Sn diffusion from the FTO substrate towards the surface, and the more defined morphology

**Table 2. Surface composition of the different samples estimated by XPS**

| Sample                      | 800C | Ti600PA | Ti800PA |
|-----------------------------|------|---------|---------|
| O <sub>lattice</sub> (at.%) | 62.4 | 65.3    | 62.7    |
| Fe (at.%)                   | 28.3 | 25.5    | 35.5    |
| Sn (at.%)                   | 8.3  | <LoD    | 1.2     |
| Ti (at.%)                   | --   | 9.2     | 1.4     |

XPS: X-ray photoelectron spectroscopy; LoD: limit of detection.

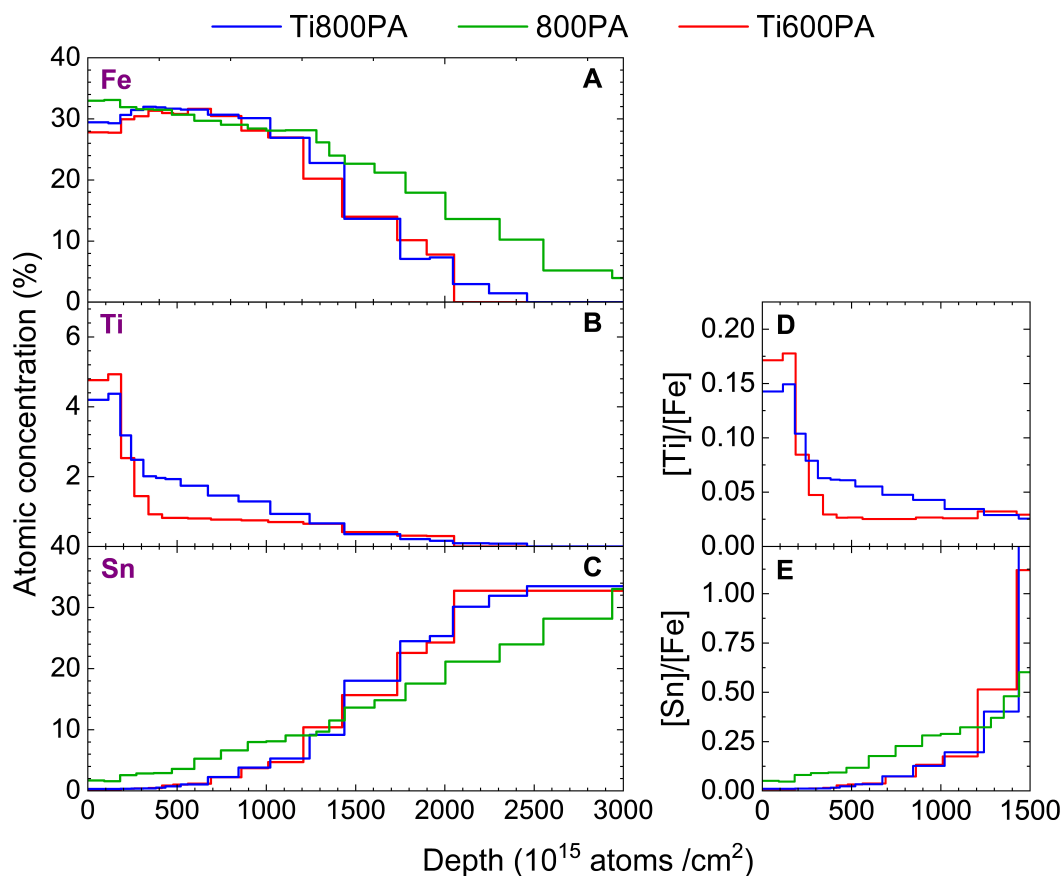
and ordering of the NWs for the undoped samples. Consequently, not all the incident ions pass through the same amount of hematite, leading to non-uniform energy loss among the ions that reach the FTO layer.

Figure 4 shows the depth concentration profile of Fe, Ti, and Sn and the variation of the [Sn]/[Fe] and [Ti]/[Fe] ratios with depth. It is observed that the Sn concentration is much higher on surface in the undoped sample than in the doped ones, which is consistent with the more defined morphology and ordering of the NWs for the undoped samples observed by SEM. This can lead to the hematite thickness being very low in certain parts of the sample, favoring the diffusion of a greater number of Sn atoms to the surface. Also, no significant difference is observed in the Sn concentration profile for the two annealing conditions applied to Ti-doped samples.

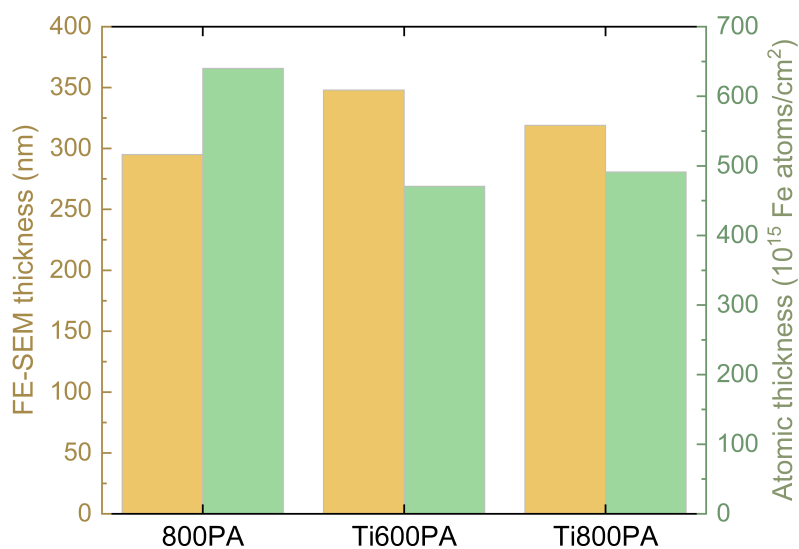
In relation to the Ti concentration, it is observed that, although in both cases the [Ti]/[Fe] ratio is greater in the most superficial part of the sample than in bulk, the difference is greater for the 600 °C annealed sample. This is consistent with the results observed by XPS. However, the difference in Ti concentration between the two samples is small compared to the results obtained by XPS. This may be due to the surface sensitivity of XPS, which analyzes only the first 5-10 nm of the sample. In contrast, while RBS can estimate a depth profile, it is not the appropriate technique to resolve the exact concentration of the most superficial part. Additionally, Ti analysis by RBS is challenging due to its low concentration, low Rutherford cross-section compared with Fe and Sn and the partial overlap of its characteristic structure with that of Fe. Therefore, while the results provide an approximate idea of the depth distribution of the different elements in the samples, the distribution of Ti presents a higher uncertainty than for Fe and Sn.

Regarding the bulk Ti concentration, it is higher for the 800 °C annealed sample. This suggests that the nucleation and growth of FeO(OH) NWs occurs in the first stage of the synthesis process, with Ti being incorporated later, mostly on the surface. Through thermal annealing processes, the diffusion of Ti throughout the Fe<sub>2</sub>O<sub>3</sub> matrix is favored. The higher the temperature, the greater this effect is confirmed to be.

By considering the atomic thickness of each layer used in the model and the percentage of Fe in each simulated layer, it is possible to estimate the total atomic thickness of the hematite layer, measured in Fe atoms/cm<sup>2</sup>, for each sample. This is shown in Figure 5, along with the thickness measured by SEM. For the undoped sample annealed at 800 °C, the atomic thickness is approximately  $6.4 \times 10^{17}$  Fe atoms/cm<sup>2</sup>. For the doped samples, the atomic thickness is  $4.7 \times 10^{17}$  Fe atoms/cm<sup>2</sup> and  $4.9 \times 10^{17}$  Fe atoms/cm<sup>2</sup> for the one-step and two-step annealing samples, respectively. This indicates that Ti doping reduces the total amount of Fe atoms in the sample by 20%-30%. However, the thickness measured by SEM for the Ti-doped samples is slightly higher than for the undoped samples. This suggests that the volume occupied by voids inside the doped samples is higher than in the case of the undoped samples, probably leading to a larger porosity for the Ti-doped photoanodes. Additionally, the fact that the two doped samples present a similar amount of Fe



**Figure 4.** In-depth concentration profiles of (A) Fe; (B) Ti and (C) Sn obtained from the analysis of the RBS spectra, for the 800PA, Ti600PA and Ti800PA samples; (D) [Ti]/[Fe] ratio, and (E) [Sn]/[Fe] ratio depth profiles. RBS: Rutherford backscattering spectrometry.



**Figure 5.** Nanometric thickness, measured by FE-SEM, and atomic thickness, measured by RBS for the 800PA, Ti600PA and Ti800PA samples. FE-SEM: Field emission scanning electron microscopy; RBS: Rutherford backscattering spectrometry.

atoms is expected since, although the distribution of these in the sample could change during the annealing processes, their incorporation is necessarily done during the hydrothermal synthesis process.

### PEIS characterization

Figure 6A shows the Nyquist plot of the Ti600PA, 800PA and Ti800PA photoanodes. Likewise, Figure 6B shows a schematic representation of the equivalent circuit used to fit the experimental data. The circuit consists of the following elements: (1) A series resistance  $R_{Series}$  corresponding mainly to the small resistance of the FTO substrate; (2) A resistance  $R_{Bulk}$  associated with the recombination of electron-hole pairs in bulk, in parallel with a capacitance  $C_{Bulk}$  associated with charge accumulation in bulk; (3) A resistance  $R_{CT,SS}$  corresponding to the charge transfer resistance from the surface states to the electrolyte, in parallel with its corresponding capacitance  $C_{SS}$ <sup>[9,96]</sup>. Capacitors were modeled as constant phase elements (CPEs), to consider the possibility of non-uniform current distribution<sup>[51]</sup>.

Table 3 shows the values of the different elements of the equivalent circuit, for the three samples analyzed. It is observed how the  $R_{Series}$  is similar and small in all cases, of about  $R_{Series} \sim 30\text{--}60\ \Omega$ . Concerning the resistance associated with the recombination of electron-hole pairs in bulk, it is minimal for the Ti800PA sample, obtaining a value of 134  $\Omega$ , higher for the 800PA sample (516  $\Omega$ ), and maximum for the Ti600PA sample (1,111  $\Omega$ ). This indicates that both Ti doping and high-temperature annealing contribute to the reduction of the recombination of electron-hole pairs in bulk. The reduction of  $R_{bulk}$  with Ti doping is consistent with the incorporation of  $Ti^{4+}$  cations into the  $\alpha\text{-Fe}_2\text{O}_3$  structure<sup>[38,59]</sup>. Also, the reduction of  $R_{bulk}$  with increasing annealing temperature is consistent with the increase in crystallinity observed for the 800 °C annealed samples, which contributes to the reduction of non-radiative recombination at grain boundaries<sup>[97–100]</sup>, as well as to the incorporation of  $Sn^{4+}$  cations at the interface with the FTO substrate when annealing at 800 °C<sup>[11,101]</sup>. The higher conductivity of the semiconductor implies a better quantum efficiency of the photoanodes due to the fewer photogenerated electron-hole pairs that recombine<sup>[102]</sup>.

On the other hand, the resistance associated with the charge transfer between the surface states and the electrolyte  $R_{CT,SS}$  is maximum for the 800PA sample (4,080  $\Omega$ ), and lower for the Ti-doped samples (1,590  $\Omega$  and 1,192  $\Omega$  for the Ti600PA and Ti800PA samples, respectively). This suggests that the presence of  $Ti^{4+}$  ions at the surface prevents positive charge accumulation at the hematite surface and decreases the surface recombination of the trapped electrons with holes from the valence band, favoring an increment of the quantum efficiency of the system.

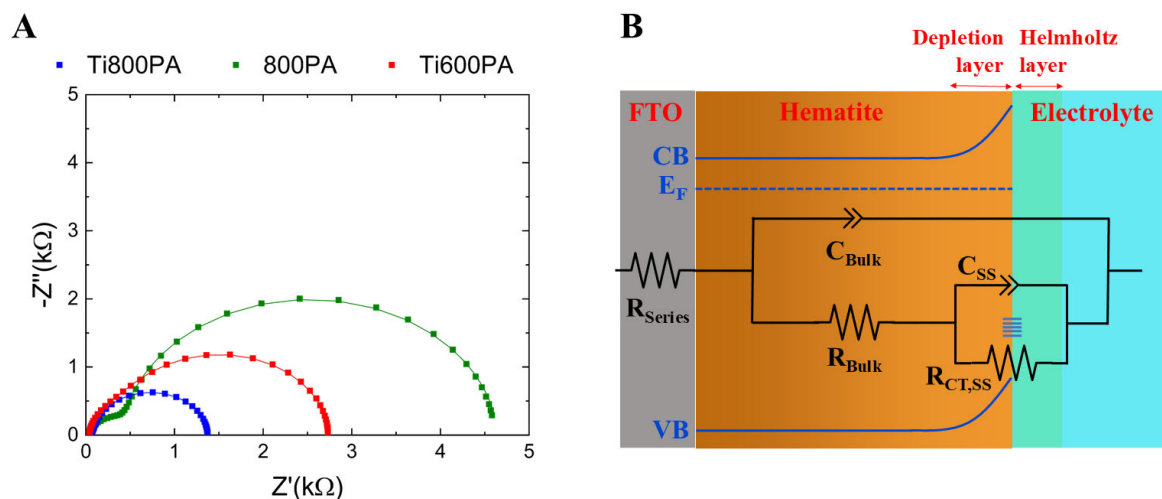
### Photoelectrochemical performance

Figure 7A shows the current  $J$ - $V$  characteristics for the synthesized  $\alpha\text{-Fe}_2\text{O}_3$  photoanodes in the dark and under simulated sunlight. For the undoped one-step annealed sample,  $J$  is very low,  $\sim 0.02\text{ mA}\cdot\text{cm}^{-2}$  (at 1.45  $V_{RHE}$ ), attributed to the low conductivity and short hole diffusion length in the obtained  $\alpha\text{-Fe}_2\text{O}_3$ <sup>[6,8,11]</sup>. The incorporation of Ti dopant allowed a  $J$  increase to 0.64  $\text{mA}\cdot\text{cm}^{-2}$  (at 1.45  $V_{RHE}$ ), about 30 times higher than that of the undoped sample. With the two-step annealing scheme, a  $J$  improvement (0.85  $\text{mA}\cdot\text{cm}^{-2}$  at 1.45  $V_{RHE}$ ) was observed for the undoped sample, while a  $J$  value of 1.27  $\text{mA}\cdot\text{cm}^{-2}$  (at 1.45  $V_{RHE}$ ) was obtained for the Ti-doped sample.

The improvement of  $J$  under sunlight illumination for the Ti-doped samples, compared to the undoped ones, is mainly due to the presence of  $Ti^{4+}$  ions on the surface, which prevent the accumulation of positive charges. Therefore, there is a decrease in the surface recombination of trapped electrons with holes in the valence band, as shown by PEIS measurements. This favors the quantum efficiency of the photoanodes. Furthermore, the reduction in bulk resistance observed, due to the presence of  $Ti^{4+}$  ions, also contributes to improving the quantum efficiency of the photoanodes.

**Table 3. Value of each element of the considered equivalent circuit, for the Ti600PA, 800PA and Ti800PA photoanodes**

| Sample                    | Ti600PA | 800PA | Ti800PA |
|---------------------------|---------|-------|---------|
| $R_{Series}$ ( $\Omega$ ) | 37      | 52    | 49      |
| $R_{Bulk}$ ( $\Omega$ )   | 1111    | 516   | 134     |
| $C_{Bulk}$ ( $\mu F$ )    | 4.1     | 4.3   | 5.0     |
| $R_{CT,SS}$ ( $\Omega$ )  | 1590    | 4080  | 1192    |
| $C_{SS}$ ( $\mu F$ )      | 7.8     | 24.4  | 5.8     |

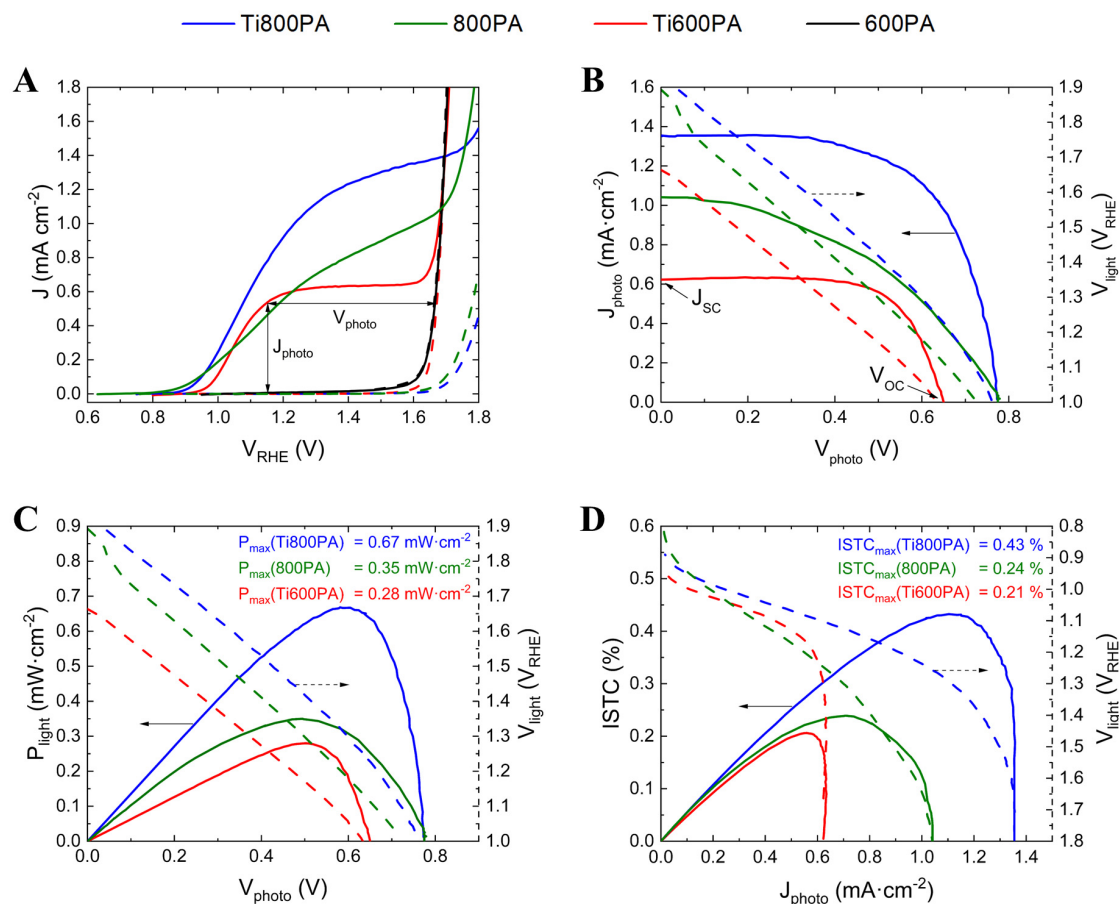

**Figure 6.** (A) Nyquist impedance plots of the Ti600PA, 800PA and Ti800PA photoanodes obtained under illumination and  $V_{RHE} = 1$  V; (B) Equivalent electrical circuit used to fit the impedance data under illumination. FTO: Fluorine-doped tin oxide.

On the other hand, the improvement of  $J$  for the samples subjected to a two-step annealing scheme, compared to the one-step annealed samples, is essentially due to a decrease in bulk recombination, as observed by PEIS measurements. This is associated, on the one hand, with an increase in crystallite size, observed by XRD, as well as with the possible incorporation of  $Sn^{4+}$  cations at the interface with the FTO substrate when annealing at 800 °C. Furthermore, the greater preferential orientation in the (110) direction of these samples also favors a greater photocurrent, since the conductivity in the [110] direction is much higher than in directions orthogonal to it<sup>[61,62]</sup>. However, despite the improvement in the photocurrent with Ti and Sn co-doping, the photocurrent value obtained is not yet comparable to reported values of up to  $\sim 6$  mA/cm<sup>2</sup><sup>[103-106]</sup>.

However, while for the undoped 800 °C annealed sample, the onset potential is  $\sim 0.90$  V<sub>RHE</sub>, an anodic shift in the onset potential is observed for the Ti-doped samples ( $\sim 0.93$  V<sub>RHE</sub> and  $\sim 0.97$  V<sub>RHE</sub> for the 800 °C and 600 °C annealed samples, respectively). This is possibly due to the additional donor levels introduced by surface doping with Ti, which affects band bending at the semiconductor-electrolyte interface<sup>[107]</sup>. Theoretical research by Toroker also indicated that a much larger overpotential was needed for water oxidation when hematite was doped with Ti on the surface<sup>[108]</sup>. Additionally, the Ti-doped samples present a step photocurrent rise. This has been reported to be associated with a reduction in recombination in surface states<sup>[109]</sup>, consistent with what was observed by PEIS.

No significant variation of  $J_{dark}$  is observed when the photoanodes are doped with Ti. However, a notable decrease in  $J_{dark}$  is observed when the samples are subjected to the two-step annealing treatment, compared





**Figure 7.** (A)  $J$ - $V$  curves for the 600PA, Ti600PA, 800PA, Ti800PA photoanodes, obtained under simulated solar illumination (solid lines), and in the dark (dashed lines); (B)  $J_{photo}$  as a function of  $V_{photo}$ ; (C) Intrinsic photovoltaic power  $P_{light}$  as a function of  $V_{photo}$ ; (D) ISTC efficiency as a function of  $J_{photo}$ . For (b-d), the potential  $V_{light}$  applied to the photoanode under simulated solar illumination to obtain each  $V_{photo}/J_{photo}$  value is shown with dashed lines.  $J$ - $V$ : Photocurrent density-voltage; ISTC: intrinsic solar-to-chemical.

to when they are subjected to the one-step annealing treatment, consistent with previous studies<sup>[9,12]</sup>.

The photocurrent  $J_{photo}$  and photopotential  $V_{photo}$  were extracted from the  $J$ - $V$  curves, as shown in Figure 7A.  $J_{photo}$  was plotted as a function of  $V_{photo}$ , so that the photocurrent density measured under short-circuit conditions  $J_{SC}$  and the open-circuit photopotential  $V_{OC}$  can be determined, as shown in Figure 7B. A  $J_{SC}$  value of 1.35 mA cm<sup>-2</sup>, 1.04 mA cm<sup>-2</sup> and 0.63 mA cm<sup>-2</sup> was obtained for the Ti800PA, 800PA and Ti600PA photoanodes, respectively, consistently with the better PEC performance of the Ti-doped samples with two-step annealing. Also, a  $V_{OC}$  value of 0.77 V, 0.77 V and 0.65 V, was obtained for the Ti800PA, 800PA and Ti600PA samples, respectively.

The intrinsic photovoltaic power  $P_{light}$  of the photoanodes as a function of  $V_{photo}$  is shown in Figure 7C. This way, the conditions for working at the MPP can be determined. For the Ti600PA and the 800PA photoanodes, maximum powers of 0.28 mW cm<sup>-2</sup> (at  $V_{light} = 1.17 V_{RHE}$ ) and 0.35 mW cm<sup>-2</sup> (at  $V_{light} = 1.31 V_{RHE}$ ) were obtained, respectively. For the Ti800PA sample, the maximum power increases substantially, obtaining a value of 0.67 mW cm<sup>-2</sup> (at  $V_{light} = 1.29 V_{RHE}$ ). This confirms the pertinence of combining Ti doping and high-temperature annealing processes to notably improve the efficiency of the PEC cell. The photocurrent  $J_{MPP}$  at the MPP was 0.56 mA cm<sup>-2</sup>, 0.71 mA cm<sup>-2</sup>, and 1.11 mA cm<sup>-2</sup> for the

Ti600PA, 800PA and Ti800PA photoanodes, respectively, while the photovoltage  $V_{MPP}$  at the *MPP* was 0.50 V, 0.49 V and 0.60 V, respectively. The *MPP* determines the optimal condition for operating the PEC cell.

The *FF* at the *MPP* was calculated for each photoanode applying eq. (3). For the Ti-doped samples, a *FF* at *MPP* of 64 % and 68 % was obtained for the photoanodes subjected to two-step and one-step annealing processes, respectively. This value is much higher than the one obtained for the undoped 800 °C annealed sample (*FF* at *MPP* of 44 %). Previous works have reported that the greater the *FF*, the greater the charge extraction compared to the recombination of electron-hole pairs<sup>[110,111]</sup>. Therefore, the fact that the Ti-doped samples present a higher *FF* indicates that charge extraction is favored by Ti-doping, consistent with the reduction of  $R_{CT,SS}$  observed by PEIS, which is associated with a reduction of the positive charge accumulation at the hematite surface.

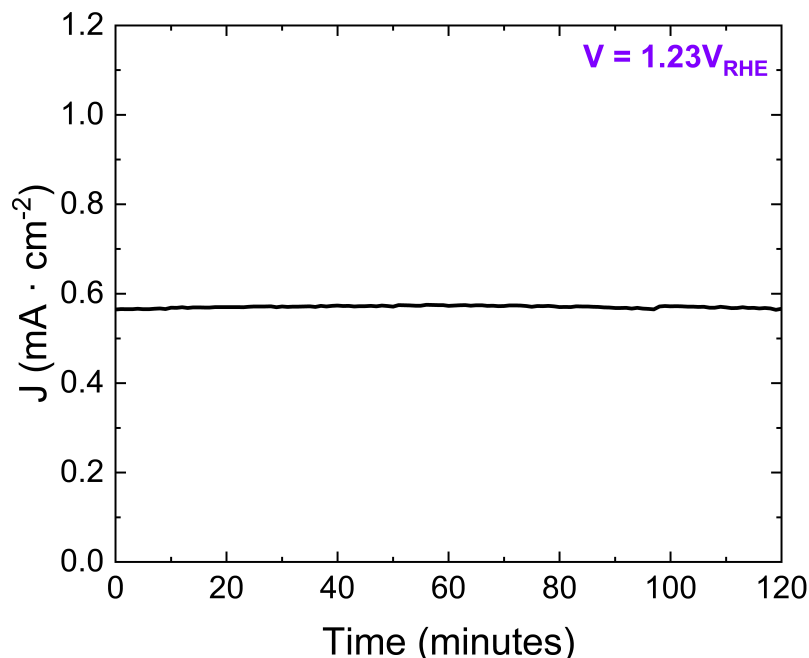
To evaluate the efficiency of the Ti-doped photoanodes in the conversion of solar into chemical energy necessary for water splitting, the *ISTC* conversion efficiency was determined by applying eq. (1). Figure 7D shows the plot of *ISTC* efficiency as a function of  $J_{photo}$ . *ISTC* maximum efficiencies of 0.21 % (at  $J_{MPP} = 0.56 \text{ mA}\cdot\text{cm}^{-2}$ ), 0.25 % (at  $J_{MPP} = 0.71 \text{ mA}\cdot\text{cm}^{-2}$ ) and 0.43 % (at  $J_{MPP} = 1.11 \text{ mA}\cdot\text{cm}^{-2}$ ) were obtained for the Ti600PA, 800PA and Ti800PA photoanodes, respectively. To obtain the same dark current values, it would be necessary to apply potentials of 1.67  $V_{RHE}$ , 1.80  $V_{RHE}$  and 1.87  $V_{RHE}$ , respectively, for the Ti600PA, 800PA and Ti800PA samples. Consequently, a simulated solar light of 0.28  $\text{mW}\cdot\text{cm}^{-2}$ , 0.35  $\text{mW}\cdot\text{cm}^{-2}$  and 0.64  $\text{mW}\cdot\text{cm}^{-2}$  for the Ti600PA, 800PA and Ti800PA photoanodes is saved.

A stability test was performed using the one-step annealed Ti-doped  $\alpha\text{-Fe}_2\text{O}_3$  sample [Figure 8].  $J$  remained stable during a continuous measurement over 2 h applying a bias potential of 1.23  $V_{RHE}$  under simulated 1-sun, showing a mean value of  $0.570 \pm 0.003 \text{ mA}\cdot\text{cm}^{-2}$ . These results confirm the good stability properties of hematite, without signs of photocorrosion<sup>[5,112–115]</sup>.

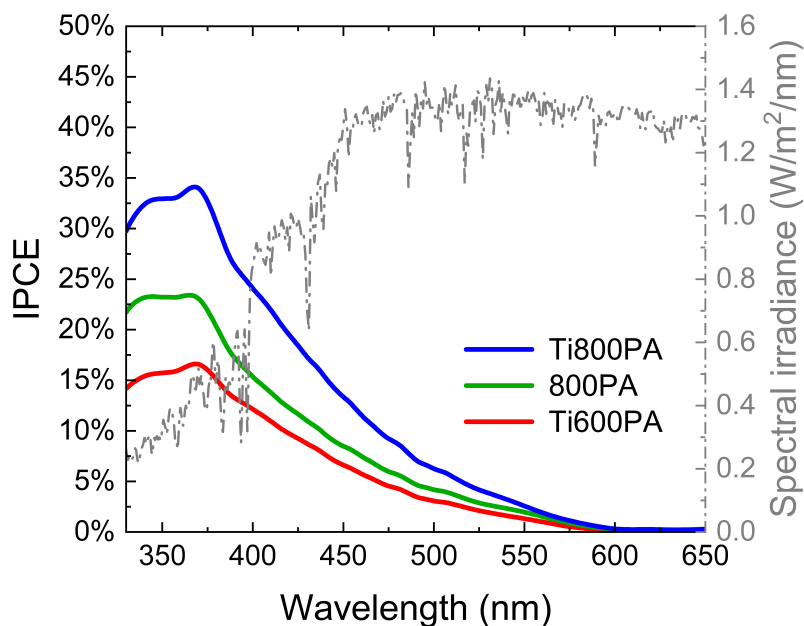
To better understand the higher photocurrent obtained for Ti-doped photoanodes, *IPCE* measurements at  $V_{RHE} = 1.45 \text{ V}$  were performed as a function of the wavelength of the incident light, as shown in Figure 9. As can be seen, the Ti800PA sample presents the highest *IPCE* at all wavelengths, with a maximum of 34%, while the 800PA and Ti600PA samples present maximum values of 23% and 17%, respectively. These results indicate a positive effect of Sn and Ti co-doping on water oxidation photoactivity. Interestingly, the Ti-doped samples exhibit a slight shoulder at approximately 370 nm. This is relevant because at wavelength values in this range the solar spectral irradiance starts to grow remarkably, so the improvement of the *IPCE* in this range contributes significantly to the enhancement of the photocurrent.

## CONCLUSIONS

Ti-doped hematite NWs were synthesized via hydrothermal synthesis, and the impact of applying a high-temperature annealing process to the photoanodes was investigated. Doping with Ti led to a notable increase in photocurrent under simulated solar illumination, rising from  $0.02 \text{ mA}\cdot\text{cm}^{-2}$  to  $0.63 \text{ mA}\cdot\text{cm}^{-2}$  (at 1.45  $V_{RHE}$ ) for the samples annealed at 600 °C, representing a 30-fold increase. This enhancement is attributed mainly to the passivation of surface states on hematite and an improvement in bulk conductivity. Additionally, when annealing was performed at 800 °C, the photocurrent increased to  $0.87 \text{ mA}\cdot\text{cm}^{-2}$  and  $1.27 \text{ mA}\cdot\text{cm}^{-2}$  (at 1.45  $V_{RHE}$ ) for the undoped and Ti-doped samples, respectively. The augmented photocurrent observed in samples subjected to high-temperature annealing is predominantly linked to the enhancement in bulk conductivity facilitated by the incorporation of  $\text{Sn}^{4+}$  ions in hematite and increased crystallinity. Notably, the Ti-doped, high-temperature annealed sample exhibited the highest photocurrent,



**Figure 8.** Photocurrent stability test for the Ti600 sample performed at 1.23  $V_{RHE}$  for 2 h.



**Figure 9.** IPCE for the 800PA, Ti600PA and Ti800PA samples, at  $V_{RHE} = 1.45$  V. In grey dashed lines, the solar radiation spectrum AM1.5 is shown. IPCE: Incident photon-to-current efficiency.

confirming the pertinence of simultaneously applying Ti doping and high-temperature annealing to improve the PEC efficiency of the hematite photoanodes. Future investigations could focus on optimizing the Ti concentration and annealing temperature to achieve the highest PEC efficiency.

By deepening our comprehension of the impact of heteroatom doping on the properties of hematite semiconductors, and the possibility of applying this together with high-temperature annealing processes, this study lays the ground for the design and enhancement of efficient materials with high specific surface areas for sustainable chemical energy conversion applications.

## DECLARATIONS

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### Availability of data and materials

The data are available upon request from the authors.

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### Conflicts of interest

All authors declared that there are no conflicts of interest.

## Ethical approval and consent to participate

Not applicable.

## Consent for publication

Not applicable.

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