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Large-area synthesis of two-dimensional MoO_{3-x} for enhanced optoelectronic applications

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Keywords: molybdenum oxide, 2D semiconductors, UV sensing, large area synthesis, vapor deposition

ABSTRACT: Two-dimensional (2D) molybdenum trioxide has been attracting research interest due to its bandgap tunability and a wide variety of desirable electronic/optoelectronic properties. However, the lack of a reproducible synthesis process for obtaining large coverage 2D MoO₃ has limited the use of this material. Here we report the synthesis of large area 2D MoO_{3-x} via physical vapor deposition, using MoO₃ powder as the precursor. The as-grown layers are directly deposited on SiO₂/Si, eliminating the necessity for any transfer process. These as-grown MoO_{3-x} layers allow for the large-scale fabrication of planar device arrays. The applicability of 2D MoO_{3-x} in optoelectronics is established via the demonstration of low-power ultraviolet (UV) sensor arrays, with rapid response times (200 μs) and responsivity up to 54.4

A.W⁻¹. At a bias voltage of 0.1 V, they are at least 400 times more power efficient than their next best contender.

1. Introduction

In recent years, ultra-thin layers of functional materials that are a few atoms thick have been investigated to reveal how their unique physical and chemical characteristics influence their electrical and optical properties.^[1] Graphene is the most investigated two-dimensional (2D) atomic crystal, since its discovery as a free-standing material.^[2] Despite the excellent optical and mechanical properties of graphene, the lack of an intrinsic electronic bandgap has led to the pursuit of other semiconducting 2D materials.^[3] As such, transition metal chalcogenides (TMDs) and oxides have been established as the alternative 2D semiconductors in place of graphene.^[3d, 4] These nanosheets, with the benefit of varying properties, are integral to the next generation of nanotechnology-enabled devices, as devices fabricated out of 2D materials demonstrate exceptional performances due to the planar confinement of charge carriers.^[3c, 5] Among them, molybdenum compounds such as molybdenum disulfide (MoS₂) and molybdenum trioxide (MoO₃) have attracted notable research efforts due to their unique properties and abundance.^[6] In particular, molybdenum trioxide (MoO₃) is a high dielectric (dielectric constant $k \sim 500$), n-type, wide-bandgap semiconductor with an easily tunable bandgap. Owing to its high dielectric constant, enhanced carrier mobilities have been observed in 2D MoO₃ in contrast to MoS₂ and other transition metal dichalcogenides.^[7] Thus, 2D MoO₃ has been established as an ideal material for electronic applications such as resistive memory devices,^[8] field effect biosensors^[6a] and electronic ink based printable transistors.^[9] MoO₃ occurs in two main crystal phases: the thermodynamically stable α -MoO₃, and the metastable β -MoO₃ that exhibits a thermal transition to α -MoO₃ above 500 °C.^[10] While α -MoO₃ is composed of stacked planar dual-layers of MoO₆ octahedra, held together by weak

Figure 1. Illustration of the crystallographic structure of (a) α -MoO₃ and (b) β -MoO₃. Red spheres indicate Mo atoms and blue spheres indicate O atoms. (c) Schematic of the experimental setup of thermal evaporation of MoO₃ powder to produce MoO_{3-x} homogenous quasi-2D layers. Blue arrows indicate the flow of Ar.

van der Waals forces along the vertical $[0\ 1\ 0]$ direction, with a layer thickness of 1.38 nm,^[10-11] β -MoO₃ occurs in cubic ReO₃ type structures where the MoO₆ octahedra are covalently bound in all three dimensions establishing a monoclinic structure (Figure 1a).^[12] Multiple synthesis techniques such as solvothermal^[13] and hydrothermal synthesis,^[14] and chemical vapor deposition (CVD)^[4b] have been proven to be effective in producing large domains of α -MoO₃ in various morphologies. However, exfoliation techniques and thermal oxidation have been the most reliable approaches to date, to realize 2D MoO₃.^[4b, 15] Exfoliation processes are either low yield or produce average quality 2D materials, while thermal oxidation does not provide control over the growth dimension due to the randomness of the nucleation sites.^[4a, 16] Until now, it remains a challenge to obtain large-area and uniform α -MoO₃ with a thickness of a few fundamental layers.^[17] Vapor deposition is a promising technique for the synthesis of large-scale MoO₃ layers of crystalline nano sheets. Ivanova, *et al.*^[18] reported CVD films with thicknesses in the range of 320-400 nm. Kalantar-zadeh, *et al.*^[16] reported the growth of bulk MoO₃ crystals where subsequent mechanical exfoliation was applied to obtain crystals of

~25 nm thickness. Although the aforementioned processes produce high-quality molybdenum oxide crystals, they are bulk crystals and the processes are not controllable with respect to the lateral dimensions as well as the thickness of the crystals formed. Recently, Molina-Mendoza *et al.* reported a centimetre-scale synthesis procedure for obtaining highly-oriented, stoichiometric α -MoO₃ crystals, using van der Waals epitaxial growth at atmospheric pressure on muscovite mica substrates.^[19] However, to realize MoO₃ based electronic/optoelectronic applications using the above mentioned synthesis processes, additional fabrication steps such as a transfer process and/or bandgap manipulation *via* the introduction of oxygen vacancies are required.^[8, 17a, 19-20] Hence, the need for a one-step, controllable process to obtain oxygen deficient, quasi-2D α -MoO₃ (MoO_{3-x}) at a large scale arises.

In this work, we report a facile and controllable synthesis process to produce uniform quasi-2D MoO_{3-x} with homogeneous coverage greater than 500×500 μm^2 in area. Subsequently, we show the applicability of this material in optoelectronics by demonstrating UV sensors arrays based on the as-grown layered MoO_{3-x} (Refer to Supporting Information S1). The devices are designed to operate at low bias voltages enabling low-power UV sensing applications. Additionally, stoichiometric MoO₃ can also be obtained by oxidizing the MoO_{3-x} layers.

2. Results and Discussion

The optimal parameters for the synthesis have been established by characterizing the composition, lateral growth homogeneity and the thickness of the as-grown material. Clustered growth of MoO_{3-x} crystals was observed at higher amounts of the precursor, which prevented the uniform growth of a 2D film. At temperatures below 500 °C, a mixed crystalline composition of α -MoO₃ and β -MoO₃ was obtained (verified via Raman spectroscopy and X-ray diffraction), whereas at temperatures above 580 °C, no significant growth was observed.

Figure 2. (a) Optical and (b) AFM images of the as grown MoO_{3-x}. The corresponding cross-sectional profile indicates the step thickness of MoO_{3-x} (c) Spatial Raman map of Si peak intensity acquired on the area shown in (a). (d) Long exposure Raman spectra obtained at the spots indicated in (a).

535 °C proved to be the ideal temperature for crystalline growth of reduced α -MoO₃. The chamber pressure was observed to influence the thickness and horizontal growth homogeneity over the substrate. Repeatability and growth consistency were achieved at a chamber pressure of 310 mTorr, which was at the vapour pressure of the precursor and the carrier gas.^[21] Further increase in the chamber pressure typically resulted in clustered growth of bulk MoO₃.

Optical microscopy revealed uniform coverage of up to 0.5×0.5 mm (Figure 1b inset) achieved on the substrates at the optimal growth conditions mentioned in the experimental section. The coverage and uniformity were also observed to be highly dependent on the cleanliness of the substrates. Figure 2a shows a magnified optical image of the as-grown MoO_{3-x} layers with an intentional fissure, revealing the substrate. Atomic force microscopy (AFM) was conducted on the area shown in Figure 2a to assess the thickness and the roughness of the as-grown MoO_{3-x} . Figure 2b shows the AFM topographic image of the area, and its corresponding cross-sectional profile. The average thickness and the roughness of the as-grown MoO_{3-x} were measured to be 8 nm (over multiple measurements at a confidence interval of 95%) and ± 0.75 nm respectively, whereas this roughness variation is within the range of a unit thickness of MoO_3 . To assess the lateral coverage of the as-grown film, spatial Raman peak intensity mapping was carried out. Figure 2c illustrates the map of Si peak intensity in the growth area of Figure 2a, indicating a homogeneous growth of MoO_{3-x} achieved over the substrate with no defects and voids. Considering the relatively low intensity of Raman peaks of the sub-10 nm thick MoO_{3-x} , a peak intensity map of SiO_2/Si was utilised to clearly identify the homogeneity of the as-grown material. Long exposure Raman spectroscopy was performed on the as-grown layers of MoO_{3-x} to assess the quality and composition of the material (Figure 2d). Dominant Raman peaks corresponding to non-stoichiometric MoO_{3-x} were observed at 126, 203, 228, 346, 361, 737 cm^{-1} and they are well in agreement with literature.^[4a, 7] Peaks indicative of the sole presence of SiO_2/Si were observed throughout the area of the fissure. Based on these results, it can be concluded that a homogeneous growth of ~ 8 nm MoO_{3-x} was obtained. Homogeneity was further confirmed in an area greater than $500 \times 500 \mu\text{m}^2$ via statistical analysis of MoO_{3-x} Raman peak intensity/shift (Refer to Supporting Information S2).

X-ray diffraction was performed to investigate the crystallinity of the as-grown MoO_{3-x} and the corresponding diffraction patterns are shown in Figure 3a. The diffraction peaks can be indexed

Figure 3. (a) X-ray diffraction patterns (b) Cross-sectional TEM image of CVD grown MoO_{3-x} . Insets showing HRTEM image and corresponding FFT from the same area of as-grown MoO_{3-x} layers. X-ray photoemission spectra of (c) Mo 3d and (d) O 1s core level of the as-grown MoO_{3-x} layers.

to MoO_{3-x} and the strong intensity of $(0k0)$ peaks (where $k = 2, 4, 6$), reveal a layered crystal structure.^[22] The above observation reveals a predominant quasi-two-dimensional growth of the crystallites along the basal planes of $\alpha\text{-MoO}_3$.^[11a] Calculated crystalline domain size from Scherrer's analysis(ref) is show to be 56.5 nm (for Scherrer calculation refer to Supporting Information S3). In order to gain further insight into the atomic structure of the as-grown MoO_{3-x} , high resolution transmission electron microscopy (HRTEM) was conducted. Figure 3b shows a TEM micrograph of a cross-section of the MoO_{3-x} film (with a scanning electron micrograph

of the as-prepared sample shown in Supporting Information S4). At the upper left edge we can see the amorphous SiO₂ substrate followed by MoO₃ crystal and a Pt protective layer on the lower right edge. No interfacial amorphous MoO_x oxide is present. The crystalline domain of a single orientation is free of defects. The annotation shows the planar spacing of 0.18 nm from the (002) plane of the α -MoO_{3-x} consistent with the XRD results. The inset Fast Fourier Transform (FFT) can be indexed to the (002) plane of orthorhombic MoO_{3-x}. As such, the TEM results along with the XRD patterns confirm the crystalline nature of the as-grown material. X-ray photoemission spectroscopy (XPS) measurements were also carried out to investigate the chemical composition and the valence states of Mo. Figure 3c and 3d show the Mo 3d and O 1s core level XPS spectra obtained from the as-synthesized material, respectively. The Mo 3d core-level spectrum could be deconvoluted showing characteristic 3d_{5/2} and 3d_{3/2} splitting components (spin-orbit splitting ~3.2 eV) for two oxidation states of Mo. The intense high binding energy (BE) doublet peak at 233.2 eV (3d_{5/2}) and 236.4 eV (3d_{3/2}) corresponds to Mo⁶⁺ oxidation state while the low intensity doublet peak at 230.3 eV (3d_{5/2}) and 233.5 eV (3d_{5/2}) corresponds to Mo⁴⁺ oxidation states. These peak positions are in good agreement with the literature.^[23] The absence of signals from molybdenum metal at BE of 228 eV and Mo⁵⁺ at BE of 231.8 eV confirms that the oxide primarily consists of Mo in 6+ oxidation state and an additional reduced 4+ oxidation state.^[24] In the current case, the fractional amounts of Mo⁶⁺ and Mo⁴⁺ are 76% and 24%, respectively, which indicates that the material predominately contains Mo⁶⁺ oxidation state. The presence of the reduced oxidation state in the current case is unsurprising as it has been previously reported that oxygen deficiencies can be formed during the thermal evaporation process. The O 1s core level spectrum could also be deconvoluted wherein an intense peak at BE 531.01 eV corresponding to the oxide layer and a lower BE peak at 532.8 eV can be assigned to surface contamination (typically from either C–O or O–C=O bonds).^[25] Taking into consideration only the peak from the oxide component, the surface stoichiometry of the oxide was calculated. This was performed by first calculating the peak

1 areas of Mo^{6+} and Mo^{5+} as well as the O 1s peak corresponding to the oxide. The O/Mo ratio
2 was equal to 2.3 suggesting the formation of MoO_3 with the presence of oxygen vacancies. This
3 supports the presence of Mo^{4+} observed in addition to the predominant Mo^{6+} in the sample.
4 In addition to understanding the quality of the as-synthesized MoO_{3-x} , another essential step is
5 to practically assess the viability of this material in optoelectronic applications. For this
6 purpose, two-terminal devices were fabricated using the as-grown MoO_{3-x} as the sensing layer.
7 The photoresponse of MoO_{3-x} was studied *via* monitoring the bias current (at a low bias voltage
8 of 0.1 V) during sequential exposures to various illumination wavelengths. Figure 4a shows the
9 time-resolved photoresponse of MoO_{3-x} when subjected to different wavelengths. The average
10 photocurrent (49 nA) was observed at ultraviolet irradiation (UV, 365 nm, 5 mW.cm^{-2}). In
11 comparison to UV, the device shows significantly lower photoresponse to the other
12 wavelengths (280, 455 and 565 nm) under the same power density. This is attributed to the wide
13 bandgap and the optical absorption of MoO_{3-x} , which restricts the absorbed photons to minimum
14 energy of about 2.72 eV in order to generate photoexcited carriers. This can be confirmed by
15 the absorption spectrum in the UV-Vis range (Figure 4b), which indicates a distinct absorption
16 peak at approximately 365 nm. To confirm the selectivity of the response, the photoresponse
17 measurements were conducted >10 times over multiple devices and found to be responsive to
18 365 nm. Figure 4b shows the trend in photoresponse (where selectivity is in alignment with
19 optical absorption) with the observed values at a confidence level of 95%.

Figure 4. (a) Time-resolved step response to multiple wavelengths (b) Selectivity observed via normalized photoresponse of the as-grown MoO_{3-x} towards different illumination wavelengths over multiple devices/measurements. Error bars in (b) represent a confidence interval of 95%. Normalized photoresponse of the as-grown few-layer MoO_{3-x} (c) the rise and fall of the photocurrent in a single pulse response cycle ($V_{\text{bias}} = 100 \text{ mV}$ and $P = 5 \text{ mW.cm}^{-2}$) (d) dependence of the photocurrent on illumination power density (e) photocurrent response during 2000 cycles of operation at $V_{\text{bias}} = 100 \text{ mV}$ and $P = 5 \text{ mW.cm}^{-2}$ (f) output photocurrent response, when subjected to varying illumination pulse frequencies.

Bulk $\alpha\text{-MoO}_3$ exhibits an indirect bandgap of $\sim 3.2 \text{ eV}$ while non-stoichiometric MoO_{3-x} shows a semiconducting behaviour with a lower bandgap.^[7] When the oxide is partially reduced, intermediate gap states are created near the bottom of the conduction band and, therefore, MoO_{3-x} exhibits a lower bandgap. Such MoO_{3-x} layers can accommodate a large number of oxygen vacancies, resulting in partially reduced Mo^{5+} states. The partially reduced cation sites also act as traps for electrons, hence facilitating photogenerated charge separation.^[26] Upon illumination, electrons in the gap states are excited to the conduction band contributing to

photoconductivity. The amount of power absorbed by the material is related to the absorption coefficient at a constant incident photon energy.^[27] Considering the absorbance peak of MoO_{3-x} is at 365 nm (Figure 4b), the highest amount of energy is absorbed by the material at this wavelength, resulting in the most substantial response, observed at 365 nm. On the other hand, there is a low sensitivity for other wavelengths (e.g. 280, 455 and 565 nm), due to the weak absorption tail of MoO_{3-x} in this wavelength range. Based on the high selectivity, the devices were assessed for their capability of sensing UV (365 nm) irradiation. Figure 4c shows the typical time-resolved response of the sensor when subjected to pulsed illumination at 365 nm. The time required for the current to increase from 5% to 95% of the peak value or *vice versa*, defines the response and recovery time, respectively. Ultra-fast response and recovery times of 200 μ s were observed when subjected to periodic illumination. To the best of our knowledge, these are the best values reported for MoO_{3-x} based UV sensors, to date.^[17a, 19-20] The photoresponse of the MoO_{3-x} based sensors depend on the illumination power as well. Figure 4d shows a gradual increase of photocurrent with illumination power density, (365 nm illumination with a power density of 0.25 to 5 mW.cm⁻² at a fixed bias of 0.1 V). The relationship between the photocurrent (I) and light intensity (P) can be described by a simple power law: $I = aP^\theta$, where a is a proportionality constant associated with the light wavelength, and exponent θ is a constant that relates device response to light intensity. By fitting this formula to the experimental data, we obtained $\theta = 0.997$ for the 365 nm light.^[28] The non-integer exponent can be regarded as a consequence of the complex process of electron-hole generation, trapping and recombination within MoO_{3-x}. Since the photogenerated carriers are proportional to the absorbed photon flux,^[29] photocurrents of the devices were observed to proportionally increase with power density. To assess the cyclic switching stability, the devices were subjected to high-frequency pulsed UV illumination. Figure 4e indicates that the responses did not deteriorate even after 2000 cycles, showing a high level of cyclic switching endurance. The frequency response of the devices was characterized by measuring the output current when

Table 1. Performance comparison of other planar material-based photodetectors

Material	Thickness	Bias voltage	Responsivity	Response time	Wavelength
This work	8 nm	0.1 V	54.4 A.W ⁻¹	200 μ s	365 nm
GaS ^[30]	5 nm	2.0 V	4.2 A.W ⁻¹	30 ms	254 nm
GaSe ^[31]	4.2 nm	5.0 V	2.8 A.W ⁻¹	20 ms	254 nm
MoO ₃ nanosheets ^[32]	78.4 nm	10.0 V	524 mA.W ⁻¹	0.94 s	380 nm
WS ₂ ^[33]	25.2 nm	5.0 V	53.3 A.W ⁻¹	---	365 nm
MoO ₃ ^[19]	4 nm	0.5 V	30 mA.W ⁻¹	20 s	405 nm
α -MoO ₃ ^[17a]	Single crystal	1.0 V	65.6 A.W ⁻¹	2 s	405 nm

subjected to illumination pulses of varying frequencies (Figure 4f). No significant decay in the output current was observed in the range of 1 Hz to 1 kHz. The variation was less than 3% within the measured frequency range, where the upper limit of the frequency was limited by the measurement instrument capabilities. Table 1 presents a summary of device performance parameters based on other well-studied semiconductor-based photodetectors, in the UV range. As shown, the responsivities of different photodetectors range from 30 mA.W⁻¹ to 65.6 A.W⁻¹, while the response time ranges from 20 ms to 20 s. In our work, the response time is observed to be 200 μ s which is orders of magnitude superior to other reported devices. Our sensors show an average responsivity of 54.4 A.W⁻¹ (for responsivity calculation refer to Supporting Information S5), which is a significant enhancement considering the low bias voltage of 0.1 V. Typical operating bias voltages range from 2 – 10 V, for TMDs and metal oxides, which leads to poor energy efficiency. It can also be seen that stoichiometric 2D MoO₃ has an extremely slow response time of 20 s.^[19] Owing to the sub-stoichiometry, quasi-2D nature, and MoO_{3-x} being an established charge transfer layer^[34], our devices are able to respond at low bias conditions rendering them at least 400 times power efficient than the next best contender ($P \propto V^2$).^[29, 35] Refer to Supporting Information S6 for I-V characteristics of a typical device. The enhancement in the absorbance due to the optical cavity effect is yet to be assessed via numerical modelling and to be experimentally verified, in our future work.

3. Conclusion

In summary, uniform large-area MoO_{3-x} layers were directly grown on SiO_2/Si substrates using a low-pressure physical vapor deposition (PVD) process. The synthesized layers display an average thickness of 8 nm (± 0.75 nm roughness), and the growth was observed to be highly crystalline. The thickness and the lateral coverage were dependent on the chamber pressure, whereas the crystallinity was reliant on the growth temperature. Based on multiple trials, optimal growth conditions were established. These as-grown MoO_{3-x} layers have a lower bandgap than bulk MoO_3 and show a large enhancement in photocurrent in the presence of 365 nm UV illumination. The frequency response of the UV sensors shows significant photocurrent stability with only a subtle variation of 1.1% within the range of 1 to 1000 Hz. The devices also demonstrate high cyclic repeatability (> 2000 cycles) over multiple measurements. The devices at a low bias (0.1 V), exhibits a linear photoresponse with power density and an enhanced responsivity of 54.4 A.W^{-1} , at a notable response time of 200 μs . As such, the results highlight the viability of MoO_{3-x} in large-scale, low power and high-performance optoelectronic applications.

4. Experimental Section

Material synthesis: The experimental setup is illustrated in Figure 1b. The growth of the MoO_{3-x} layers was carried out in a low-pressure, three zoned horizontal split tube furnace (Labec HTFS80) in a 70 mm diameter quartz tube. Prior to the PVD process, 300 nm SiO_2/Si substrates were sequentially cleaned using acetone, IPA, and DI water in an ultrasonic bath for 200 s each. Subsequently, the samples were blow dried (N_2), and O_2 plasma (RF power of 30 W) treated for 300 s. 35 mg MoO_3 powder (Sigma-Aldrich 98%) was placed in a quartz boat in the center of the evaporation zone T_1 , and the growth substrate (2×2 cm) was placed upstream relative to

the gas flow direction, at the center of growth zone T_3 . The quartz tube was evacuated (240 mTorr) and the optimal deposition pressure of 310 mTorr was achieved at a steady flow of 100 sccm Ar. The zone temperatures T_1 , T_2 and T_3 , were set to 785, 630 and 530 °C respectively and were achieved at a ramp-up rate of 25°C/min. The PVD system was held at the aforementioned temperatures for a period of 45 min for the deposition to occur and was naturally cooled down to room temperature. The aforementioned optimal parameters (for obtaining large-area, few-layered growth) were achieved after several trials of applying various conditions. The pressure was maintained at 310 mTorr throughout the synthesis process.

Characterization: The experimental Atomic force microscopy (AFM) was performed to determine the thickness of the as-grown layers and to investigate the morphology of the grown MoO_{3-x} . AFM imaging was conducted on a Dimension Icon-Bruker instrument in the contact mode. Spatial Raman peak intensity mapping was acquired using a Horiba LabRAM HR Evolution Raman spectrometer (Laser excitation at 532 nm with a power of 9 mW). Spectra were collected through a 50× Leica objective and recorded with an 1800 lines mm^{-1} grating that provided a lateral resolution of $\sim 0.5 \mu\text{m}$ with a 30 s exposure time.

The X-ray diffractograms (XRD) were collected using a Bruker D4 Endeavor XRD instrument with monochromatic $\text{Cu K}\alpha$ as the radiation source, within an angular range of 2θ from 20° to 60° and X-ray photoelectron (XPS) measurements were carried out using a Thermo K-Alpha XPS instrument using an $\text{Al K}\alpha$ X-ray source (photon energy of 1486.7 eV) at a pressure of 1×10^{-9} Torr ($1 \text{ Torr} = 1.333 \times 10^2 \text{ Pa}$). The Mo 3d and O 1s core level spectra were recorded at a pass angle of 90° with an overall resolution of 0.1 eV. The XPS spectra were background corrected using Shirley algorithm and the chemically distinct species were resolved using standard Gaussian-Lorentzian function. The spectra were aligned using the adventitious carbon binding energy of 285 eV. Transmission Electron Microscopy (TEM) was performed using a JEOL 2100F operating at 200kV equipped with a Gatan OneView camera. Samples were

prepared using Focused Ion Beam (FIB) with a Ga liquid metal ion source. Samples were thinned to ~100 nm at 30 KV and further polished at 5 kV to removed amorphous ion beam damaged area.

Device fabrication: The experimental Photodetector devices using the as-grown MoO_{3-x} as the sensitive layer were fabricated in a two-terminal configuration *via* standard photolithography. Metal electrodes were realized using electron-beam evaporation and lift-off processes. The electrodes were composed of 80 nm of Au on 10 nm of Cr adhesion layer. Commercial high power light-emitting diodes (Thorlabs, Inc.) with various wavelengths (280, 365, 455, and 565 nm) were used as illumination sources. The illumination power ($5 \pm 0.02 \text{ mW.cm}^{-2}$) was calibrated using a commercial UV-enhanced silicon photodetector (Newport Corporation). The electrical measurements were performed using an Agilent 2912A source meter and a Keithley 4200SCS semiconductor characterization system. UV-Vis absorption spectra were collected using a CRAIC 20/30 microspectrophotometer.

Supporting Information

Supporting Information is available from the author.

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