Controllable P-type Doping of 2D WSe$_2$ via Vanadium Substitution

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Abstract

Scalable substitutional doping of two-dimensional (2D) transition metal dichalcogenides (TMDCs) is a prerequisite to developing next-generation logic and memory devices based on 2D materials. To date, doping efforts are still nascent. Here, we report scalable growth and vanadium (V) doping of 2D WSe$_2$ at front-end-of-line (FEOL) and back-end-of-line (BEOL) compatible temperatures at 800 °C and 400 °C. A combination of experimental and theoretical studies confirm that vanadium atoms substitutionally replace tungsten in WSe$_2$, which results in p-type doping via the introduction of discrete defect levels that lie close to the valence band maxima. The p-type nature of the V dopants is further verified by constructed field-effect transistors, where hole conduction becomes dominant with increasing vanadium concentration. Hence, our study presents a method to have precise control over the density of intentionally introduced impurities, which will be indispensable in the production of electronic-grade wafer-scale extrinsic 2D semiconductors.
Introduction

Doping of transition metal dichalcogenides (TMDCs) is the major prerequisite to realize next-generation logic transistors with ultimate gate length scaling. Both intrinsic and extrinsic TMDCs have been actively explored as a channel material at front-end-of-line (FEOL) and as a diffusion barrier, liner, and thin-film transistor (TFT) at back-end-of-line (BEOL) for advanced semiconductor technology nodes. To date, wafer-scale synthesis of near electronic-grade intrinsic TMDCs has been successfully realized at FEOL (>800 °C) and BEOL (<500 °C) compatible temperatures whereas the same progress does not hold true for scalable doping. Various doping efforts have been developed to tune the electrical conductivities of TMDCs using surface charge transfer doping, electrostatic doping, intercalation, and substitutional doping. Even though “proof-of-concept” devices have been demonstrated on doped TMDCs, uniform distribution and tight control of the impurity density over the large scale, as well as compatibility of the introduced methods with the state-of-the-art Si CMOS 300 mm process lines, still remain challenging. Thus, scalable doping of TMDCs over a large temperature window with accurate control over the doping concentration is urgently needed.

Here, we report substitutional vanadium doping of WSe$_2$ films realized via MOCVD at BEOL and FEOL compatible temperatures using W(CO)$_6$, H$_2$Se, and (C$_5$H$_5$)$_2$V as metal, chalcogen, and the dopant sources, respectively. Accurate control over the partial pressure of the reactants enables uniform dopant distribution over the surface with a predetermined vanadium concentration, verified by high-resolution X-ray photoelectron spectroscopy (XPS). Furthermore, room temperature Raman and photoluminescence (PL) spectroscopy studies reveal a strong dopant concentration dependence of the spectra where defect activated modes increase in intensity and positive trions are formed that completely quench the PL spectra. Atomic-scale characterization techniques such as high-resolution transmission electron microscopy (HRTEM), scanning tunneling microscopy and spectroscopy (STM/STS), and first-principles density functional theory (DFT) calculations confirm that vanadium atoms substitutionally replace W atoms in the WSe$_2$ lattice and introduce multiple defect states that are in close proximity to the valence band edge. The p-type nature of the vanadium dopants is further confirmed via back-gated field-effect transistors that exhibit an enhanced p-branch (hole) current and a positive threshold shift. Our systematic study provides foundational knowledge in large-scale synthesis of extrinsic 2D TMDCs with precisely controlled dopant concentrations that will be a major step towards successful integration with Si CMOS at FEOL and BEOL.

Results and Discussion:

Pristine and V-doped FEOL-compatible WSe$_2$ is synthesized on c-plane sapphire substrates at 800 °C and 700 Torr using W(CO)$_6$, V(C$_5$H$_5$)$_2$, and H$_2$Se, respectively. The obtained films are fully coalesced and monolayer in nature with some bilayer islands on the top where the intentional introduction of vanadium atoms has a direct impact on the surface morphology (Figure 1a, b). Independent of the dopant concentration, the shape of the domains (size 800-900 nm) in extrinsic films starts to deviate from the triangular shape observed in intrinsic WSe$_2$, becoming truncated triangles and hexagons (Figure S1a-d). Furthermore, an increase in density of bilayer islands occurs in extrinsic films with increasing dopant concentration (Figure S1c, d), suggesting that vanadium adatoms serve as secondary nucleation centers on the surface. In parallel, BEOL-compatible 2D-WSe$_2$ is obtained on SiO$_2$/Si substrates at 400 °C and 700 Torr, with all films being coalesced, multilayer, and polycrystalline with an average domain size of 100 nm and thickness of 2.11 nm, corresponding to 3-4 layers (Figure S2a-d and S3a, b). Moreover, chemical composition analysis is carried out using high-resolution XPS, providing evidence that that V(C$_5$H$_5$)$_2$ decomposes at FEOL and BEOL temperatures and incorporates V into the WSe$_2$ lattice at either temperature. Both intrinsic and extrinsic films exhibit characteristic W 4f$_{7/2}$, W 4f$_{5/2}$, and W 5p$_{3/2}$ peaks at 32.6 eV, 34.8 eV, 38.1 eV and Se 3d$_{5/2}$ and Se 3d$_{3/2}$ peaks at 54.8 eV and 55.7 eV, respectively (Figure 1c and S4a) agreeing with previous reports. As the vanadium concentration increases, associated V 2p$_{3/2}$ and
V $2p_{3/2}$ XPS peaks are detectable at 513.7 eV and 521.7 eV, respectively, indicating successful substitutional doping of V at the W site of the WSe$_2$ lattice (Figure 1d and S4b). Moreover, a peak at ~517.0 eV is observed that corresponds to V-O bonding, further suggesting vulnerability of extrinsic films towards oxidation (Figure 1d and Figure S4b). The similar characteristic XPS W 4f, Se 3d and V 2p peaks are observed on BEOL V-WSe$_2$ on SiO$_2$/Si, where the dopant concentration varies as a function of the vanadium precursor flow rate (Figure S5a, b).

Figure 1. Surface morphology, chemical composition, and optical properties of pristine and V-doped WSe$_2$ films on sapphire substrates. AFM images of (a) pristine and (b) 1.9% vanadium doped WSe$_2$ films demonstrating a clear impact of the dopants on the surface morphology (1μm scale bar). (c, d) High-resolution XPS analysis further confirms the presence of vanadium atoms in the lattice where V 2p spectra are found at 513.7 eV, 517.0 eV, and 521.7 eV. (e, f) Corresponding Raman spectra display increased density of defect activated modes whereas PL spectra start to redshift and are completely quenched (1.9 % V-WSe$_2$) as a function of vanadium concentration, respectively.

Vanadium doping leads to enhancement in Raman defect modes and quenching of WSe$_2$ photoluminescence. Intrinsic and extrinsic FEOL compatible films exhibit peaks at 249.3 cm$^{-1}$ and 260.6 cm$^{-1}$ corresponding to A+E and 2 LA(M) characteristic in and out-of-plane Raman active modes of 2D WSe$_2$. Intensified defect activated modes (ZA(M) (109.2 cm$^{-1}$) and LA(M) (128.6 cm$^{-1}$)) as vanadium is introduced are the direct result of the increased structural disorder in the WSe$_2$ lattice and a strong function of the concentration. Corresponding PL spectra exhibits similar dopant concentration dependence where intrinsic WSe$_2$ displays an optical band gap of 1.65 eV, whereas the PL peak in V-WSe$_2$ is redshifted by 40 meV (0.6% V-WSe$_2$) and 80 meV (1.1% V-WSe$_2$), and PL intensity is reduced. As dopant concentration increases to 1.9%, there is a complete quench in the PL. The observed PL quenching may be due to an increased density of positive trions, since vanadium is expected to be a p-type dopant in WSe$_2$ and donates an extra hole to the system. Charged trions, unlike neutral excitons experience stronger electrostatic interaction with the dopant impurities resulting in non-radiative recombination that completely quenches the PL intensity for vanadium concentrations >1.9%. As expected, BEOL V-WSe$_2$ films display similar Raman and PL properties (Figure S5c, d) as a function of vanadium concentration where the host lattice is perturbed after the dopant incorporation.
Atomic resolution high-angle annular dark-field scanning transmission electron microscopy (HAADF – STEM) confirms the presence of vanadium atoms in the lattice (Figure 2a). The technique relies on contrast difference based on atomic number\(^{[1]}\), where vanadium atoms are expected to exhibit less intensity compared to W atoms owing to their smaller atomic number in the periodic table. The dopants (green dashed circles) are uniformly distributed throughout the WSe\(_2\) lattice. The extrinsic films also contain other types of point defects such as single Se vacancies (red dashed circles) and double Se vacancies (yellow dashed circles) that are commonly observed in the WSe\(_2\) lattice\(^{[2]}\) (Figure 2a). Moreover, selected area electron diffraction (SAED) pattern exhibits single crystal signature with hexagonal symmetry (Figure 2a inset) indicating oriented growth of the film on the sapphire substrate. An intensity line profile obtained along an armchair direction containing one such V-dopant confirms the substitutional nature of the V dopant atom at the W site of the lattice (Figure 2b and inset) supporting our previous observations via XPS, Raman and PL. This is evident from the low intensity peak corresponding to the V dopant atom whereas an adatom results in a higher intensity peak due to the combined intensity from W and V atoms. Furthermore, STEM-energy dispersive spectroscopy (EDS) mapping (Figure S6a-e) underlines the uniform distribution of constituent elements in the film.

**Figure 2.** Atomic resolution imaging showing the possible position of vanadium dopants in the WSe\(_2\) lattice, and the corresponding band structure and projected density of states calculated by first principles density functional theory (DFT). (a) HAADF-STEM image showing V dopants (green dashed circles), single Se vacancies (red dashed circles) and double Se vacancies (yellow dashed circles) and SEAD pattern (inset) displaying single crystal nature of the films with hexagonal symmetry. (b) Higher magnification HAADF-STEM image showing one V dopant atom where the blue box indicates an armchair direction along which the intensity profile is obtained (inset) highlighting the substitutional doping of the V dopant atom. (c) Calculated band structure and projected density of states for the ionized V dopant on a W site, including spin–orbit coupling. The density of states is projected onto the \(d\)-orbitals of the V dopant atom, with the gray shading indicating the density of states for pristine WSe\(_2\), for comparison. The most relevant defect state introduced by the V dopant near the top of the valence band is highlighted in the orange box.

First-principles DFT calculations (see Materials and Methods) confirm that the W site is the most energetically favorable site in the WSe\(_2\) lattice for the V dopant to occupy (Table S1), in agreement with the experimental findings. No significant atomic relaxations or symmetry breaking of the lattice in the vicinity of the dopant is observed in the calculations. The DFT-computed electronic band structure and density of states (Figure 2c) indicate that the V dopant introduces energy states (highlighted in orange) close to the top of the valence band. By projecting the density of states onto the individual \(d\)-orbitals of the V dopant atom, we can determine that this defect state originates primarily from the \(V\ 3d_{z^2}\) orbital. When in the neutral (un-ionized) state, the dopant level is only partially occupied; however, the dopant is easily
ionized, becoming negatively charged, and the fully filled dopant level falls slightly below the top of the valence band (Figure 2c) agreeing well with previous reports.[3]

In addition to the V dopant and Se vacancies, a variety of other point defects can exist in V-doped WSe$_2$. The many point defects are identified either as Re, Mo, or Cr substituting for W,[4,5] or as oxygen substituting for selenium,[6] respectively. Consequently, we assign the most common point defect, appearing as a black depression in the STM topography, as vanadium impurities. Combining STM and CO-tip noncontact atomic force microscopy (nc-AFM) reveals that V impurities substitutionally replace W atoms in the WSe$_2$ lattice (Figure 3b,c). A vanadium dopant appears as a dark depression at positive sample bias (Figure 4a, b), but as a bright protrusion at negative voltage (Fig. 3d/4b), indicating it is a negatively charged (ionized) impurity. Indeed, Kelvin probe force microscopy (KPFM) experiments (Figure 3e) corroborate V impurities to be negatively charged by the positive shift of the local contact potential difference (LCPD).[7] The STS spectra of individual, relatively isolated V dopants (Figure 4c) reveal multiple in-gap states in the occupied spectrum, close to the valence band edge, in agreement with the expected formation of p-type defect states. At positive sample bias, the conduction band onset is pushed towards higher energies in the vicinity of dopants resulting from local band bending.[8] When the dopants are very close to each other, the defects exhibit a spectrum that is shifted towards higher energies (cf. orange vs blue curve in Figure 4f), an electrostatic effect due to the negative defect charge analog to the negative energy shift observed for positively charged n-type dopants.[5] This effect can be as large as several hundred millivolts, shifting the most tightly bound defect state up to the Fermi level.

Figure 3. (a) STM overview image ($V = 1.8$ V, $I = 50$ pA) of V-doped monolayer WSe$_2$ on EG/SiC. Individual V dopants appear as a black dot at positive sample voltage. (b,c) STM and CO-tip nc-AFM image of a single V substituent at a W site. The $V_W$ position is marked by a white dot. The unit cell is indicated at the bottom left in (c) ($W =$ blue, $Se =$ orange). (d) STM image of $V_W$ at negative ($V = -1.2$ V) sample bias. (e) Kelvin probe force microscopy (KPFM) measurement across the $V_W$ defect (along arrow in d) exhibiting a positive shift of the local contact potential difference (LCPD), indicative of a negative defect charge.
Transport property measurements of the extrinsic films further confirms the p-type nature of vanadium dopants. Back-gated-field-effect transistors (BGFETs) were fabricated on an Al₂O₃/Pt/TiN/p++ Si substrate, where e-beam lithography was used to define FET structures. Transfer characteristics of pristine and lightly doped (0.6%) WSe₂ show (Figure 5 a, b) ambipolar characteristics, with both n- and p-branches present. However, increasing the doping concentration to 1.1% and 1.9% (Figure 5 c, d) lead to enhancement in the p-branch and degradation of the n-branch. Moreover, the threshold voltage increases with higher V concentrations, confirming the p-dopant nature of the vanadium atoms in the WSe₂, agreeing with STM and DFT observations. Hole mobility is extracted in the linear regime of the transfer curve using the peak transconductance method given by the equation:

$$\mu_h = \frac{dI_{DS}}{dV_{BG}} \left( \frac{W}{L_{CH}} C_{ox} V_{DS} \right)^{-1}$$

where $W$ is the width, $L_{CH}$ is the length of the channel, $C_{ox}$ is the gate capacitance per unit area given as $C_{ox} = \frac{\varepsilon_0 \varepsilon_{ox}}{t_{ox}}$ where $\varepsilon_0$ is the dielectric constant of free space, $\varepsilon_{ox}$ is the dielectric constant of Al₂O₃, and $t_{ox}$ is the thickness of the gate dielectric. The threshold voltage ($V_{th}$) was extracted using the constant current method for $I_{DS} = 100\text{nA}/\mu\text{m}$. All the constructed devices (Table S2) display ON-OFF ratio ($I_{ON}/I_{OFF}$)
exceeding \(10^6\) and mobility values similar to earlier reports\(^{[9,10]}\). Improvement in the \(I_{ON}\) for the p channel, without losing gate modulation, further highlights the effectiveness of vanadium for p-doping over other p-doping strategies\(^{[10–13]}\).

**Figure 5.** Transport properties of intrinsic and extrinsic FEOL WSe\(_2\) back-gate field-effect transistors (BGFETs) transferred on 50 nm ALD Al\(_2\)O\(_3\) substrates. (a) Transfer characteristics of the intrinsic films (\(I_{DS}\) vs \(I_{BG}\)) at a different drain voltages (\(V_{DS}\)) display expected ambipolar conduction. (b)-(d) However, electron current (n-branch) starts to degrade as a function of vanadium concentration and BGFETs are no longer ambipolar and predominantly p-type.

**Conclusion**

We have demonstrated the scalable synthesis and vanadium doping of 2D WSe\(_2\) via MOCVD at FEOL and BEOL compatible temperatures. Our experimental findings confirm the successful incorporation of the dopants by substitutionally replacing W atoms in the host lattice, where the concentration of vanadium atoms can be precisely tuned by having accurate control over the flux of the vanadium precursor. Atomic-scale characterization reveals that the dopants are uniformly distributed throughout the WSe\(_2\) lattice. Moreover, incorporated V atoms introduce discrete energy levels that overlap the top of the valence band which confirms the expected p-type nature of the dopants and agrees well with our DFT calculations and electrical tests. We believe that the method developed in the current study can help to bridge the gap between academic research and industry by enabling wafer-scale production of intrinsic and extrinsic materials at various temperatures with the highest precision which is crucial for achieving integration of 2D materials with Si CMOS.

**Experimental Section:**

*MOCVD growth of pristine and V-doped WSe\(_2\) films.* A custom-designed vertical cold wall gas-source CVD reactor is utilized to synthesize intrinsic and extrinsic WSe\(_2\) films as previously reported\(^{[5,14]}\). The tungsten hexacarbonyl (W(CO)\(_6\)) (99.99 %, Sigma-Aldrich), Bis(cyclopentadienyl)vanadium (V\(_2\)(C\(_5\)H\(_5\))) (sublimed, 95%, Strem Chemicals), and hydrogen selenide (H\(_2\)Se) (99.99% Matheson) are used as metal, dopant and chalcogen precursors, respectively in a 100% H\(_2\) ambient. The solid-state metal and dopant sources are kept inside the stainless-steel bubblers where temperature and pressure of the bubblers are constantly maintained at 25 °C and 725 Torr, and 40 °C and 725 Torr, respectively. Hydrogen selenide is supplied from a different gas manifold and all three precursors are introduced from separate lines to prevent the intermixing before reaching the reactor inlet. The three-step growth (nucleation, ripening, and lateral growth) method\(^{[15]}\) is used to grow monolayer, epitaxial intrinsic, and extrinsic FEOL WSe\(_2\) on c-plane sapphire (Cryscore Optoelectronic Ltd, 99.996%) substrates. Growth temperature, pressure, and HSe\(_2\) flow rate are kept constantly at 800 °C, 700 Torr, and 7 sccm, respectively, for all three growth steps. At the nucleation stage,
the metal and the dopant precursors are introduced simultaneously with a flow rate of 20 sccm and 60 sccm, respectively, for 2 min. At the ripening stage, W(CO)_6 and V_2(C_5H_5)_2 were switched off and formed V-WSe_2 nuclei are let anneal under H_2Se for 10 min. W is re-introduced with a constant 4.5 sccm at the lateral growth stage, whereas the dopant flow rate varied from 30 sccm to 5 sccm depending on the desired vanadium content in the films. The BEOL counterparts are synthesized with a single-step growth method on SiO_2/Si substrates (University Wafer, Silicon P/B (100)) with 300 nm wet thermal oxide where the temperature and pressure of the reaction chamber are kept at 400 °C and 700 Torr, respectively.\cite{5,14} To promote the lateral growth of the pristine and V-WSe_2 films at this kinetically limited growth regime, the precursors flow rates were adjusted accordingly to 2 sccm and 9 sccm for W(CO)_6 and H_2Se, respectively whereas the V_2(C_5H_5)_2 flow rate is varied from 10 to 2.5 sccm. Prior to the growth, sapphire and SiO_2/Si substrates are cleaned with acetone and isopropyl alcohol (IPA) in an ultrasonication bath for 10 min each followed by DI water rinse and N_2 gun dry. To further minimize the organic contamination on the surface, the substrates are cleaned with commercially available heated Piranha solution (Nanostrip, KMG Electronic Chemicals) at 90 °C for 20 min and rinsed with deionized (DI) water.

Atomic Force Microscopy (AFM). Bruker Icon I tool is used to acquire AFM data in a peak force tapping mode.

Raman and Photoluminescence (PL) spectroscopy. Raman and PL spectra of the samples are obtained by the Horiba Labram HR Evolution VIS-NIR Raman system with a 633 nm laser at 0.4 mW power and 532 nm laser at 0.4 mW power, respectively.

Scanning Electron Microscopy (SEM). Verios G4 with the accelerating voltage of 2 keV is employed to analyze the films.

X-ray Photoelectron Spectroscopy (XPS). High-resolution X-ray photoelectron spectroscopy data is obtained by a Physical Electronics Versa Probe II tool with a monochromatic Al Kα x-ray source (hν=1486.7 eV) at high vacuum (<10^-6 Torr) environment. The acquired spectra are charge corrected to C1s core level at 284.8 eV and W 4f_7/2 at 32.7 eV, respectively. U 2 Tougaard background is used for V and W whereas Iterated Shirley background is used for Se to fit the XPS spectra of the samples.

Transmission Electron Microscopy (TEM). As grown V-WSe_2 films are lifted from sapphire substrates and transferred to Quantifoil Cu TEM grids using a PMMA assisted transfer method.\cite{16} Selected area electron diffraction (SAED) patterns and scanning transmission electron microscopy-energy dispersive spectroscopy (STEM-EDS) maps are acquired using a Thermo Fisher Talos F200X instrument at 80 kV acceleration voltage. A Thermo Fisher Tian^3 G2 microscope equipped with image and probe correctors is employed at 80 kV to collect atomic resolution high angle annular dark field (HAADF)-STEM data. The images are acquired using 30 mrad semi convergence angle, ~ 50 pA screen current and 80 kV acceleration voltage to minimize the possible damage to the samples. Acquired atomic resolution images are smoothed with a 2-pixel gaussian blur filter using ImageJ.

STM, STS, nc-AFM and KPFM. V-doped WSe_2 FEOL samples were prepared ex situ on EG on SiC substrates followed by a final 450 °C anneal in ultrahigh vacuum. The measurements were performed with a commercial QPlus from Scienta Omicron operated at 5 K and at pressures below 2×10^-10 mbar. For STM, STS and KPFM the tungsten tip mounted on a QPlus tuning fork sensor was prepared on a clean Au(111) surface (sputtering: 10 min, Ar+, 1kV; annealing: 10 min, 450 °C) and confirmed to be metallic. For nc-AFM experiments, the metallic tip was modified with a single CO molecule\cite{17} that was picked up from a Au(111) surface. STM topographic measurements were taken in constant current mode with the bias voltage
given with respect to the sample. STS measurements were recorded using a lock-in amplifier at 610 Hz and a modulation amplitude of 20 meV. The nc-AFM topographies and KPFM measurements were acquired in constant height mode, while the QPlus sensor was driven at its resonance frequency (of about 25 kHz) with a constant amplitude of 70 pm. The frequency shift from resonance of the tuning fork was recorded using Omicron Matrix electronics and HF2Li PLL from Zurich Instruments.

**DFT calculations.** The formation energy and electronic structure of the substitutional V dopant was computed using DFT as implemented in the Vienna ab initio simulation package VASP.\(^{[18]}\) Calculations were performed using projector-augmented wave potentials\(^{[19,20]}\) with Perdew–Burke–Ernzerhof (PBE)\(^{[21]}\) generalized gradient approximation functionals to treat the exchange-correlation. Spin-polarized calculations with spin-orbit coupling were performed with a plane wave cutoff energy of 520 eV, Methfessel–Paxton smearing\(^{[22]}\) with a smearing energy width of 0.10 eV, and Γ-centered Monkhorst-Pack k-point meshes\(^{[23]}\) for Brillouin zone integration. The dopant was modeled using 4 × 4 and 5 × 5 supercells containing a single dopant atom each, and 20 Å vacuum spacing between layers to minimize interlayer interaction.

The formation energy \(E_f[X]\) of a neutral point defect \(X\) was determined from DFT calculations using a supercell approach following \(E_f[X] = E_{tot}[X] - E_{tot}[\text{pristine}] - \sum_i n_i \mu_i\), where \(E_{tot}[X]\) and \(E_{tot}[\text{pristine}]\) are the total DFT-computed energies of the supercell containing the defect \(X\) and the pristine supercell respectively, \(n_i\) is the number of atoms of species \(i\) added/removed, and \(\mu_i\) is the corresponding chemical potential of the species. The values reported in this work were referenced to bcc V for \(\mu_V\), and either bcc W for \(\mu_W\) (W-rich limit), or hexagonal (gray) Se for \(\mu_{Se}\) (Se-rich limit).

**Fabrication of field effect transistors.** The as-grown pristine and V-doped WSe\(_2\) films are transferred onto a 50 nm ALD Al\(_2\)O\(_3\) on Pt/TiN/p++ Si from the host c-plane sapphire substrates. Photoresist is spun and 1µm by 5µm strips of V-WSe\(_2\) are isolated from the conformal film using e-beam lithography (EBPG 5200 Vistec). After developing, a Plasma-Therm Versalock 700 inductively coupled plasma etch tool is used (SF\(_6\)/O\(_2\) 30/10 sccm gas mixture) to remove the exposed film. Photoresist is spun and source and drain contacts are defined using e-beam lithography, followed by 40 nm Ni and 30 nm Au contact metal deposition in a Temescal e-beam evaporator tool. The electrical characterization is conducted using a Keysight B1500A semiconductor parameter analyzer; the measurements are done at room temperature and at \(~10^{-5}\) Torr.

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**Conflict of Interest**

Authors declare no conflict of interest.
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