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Engineering the interface chemistry for scandium electron contacts in WSe₂ transistors and diodes

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Supplementary material for this article is available online

Abstract

Sc has been employed as an electron contact to a number of two-dimensional (2D) materials (e.g. MoS₂, black phosphorous) and has enabled, at times, the lowest electron contact resistance. However, the extremely reactive nature of Sc leads to stringent processing requirements and metastable device performance with no true understanding of how to achieve consistent, high-performance Sc contacts. In this work, WSe₂ transistors with impressive subthreshold slope (109 mV dec⁻¹) and \(I_{ON}/I_{OFF}\) (10⁶) are demonstrated without post-metallization processing by depositing Sc contacts in ultra-high vacuum (UHV) at room temperature (RT). The lowest electron Schottky barrier height (SBH) is achieved by mildly oxidizing the WSe₂ in situ before metallization, which minimizes subsequent reactions between Sc and WSe₂. Post metallization anneals in reducing environments (UHV, forming gas) degrade the \(I_{ON}/I_{OFF}\) by ~10³ and increase the subthreshold slope by a factor of 10. X-ray photoelectron spectroscopy indicates the anneals increase the electron SBH by 0.4–0.5 eV and correspondingly convert 100% of the deposited Sc contacts to intermetallic or scandium oxide. Raman spectroscopy and scanning transmission electron microscopy highlight the highly exothermic reactions between Sc and WSe₂, which consume at least one layer RT and at least three layers after the 400 °C anneals. The observed layer consumption necessitates multiple sacrificial WSe₂ layers during fabrication. Scanning tunneling microscopy/spectroscopy elucidate the enhanced local density of states below the WSe₂ Fermi level around individual Sc atoms in the WSe₂ lattice, which directly connects the scandium selenide intermetallic with the unexpectedly large electron SBH. The interface chemistry and structural properties are correlated with Sc–WSe₂ transistor and diode performance. The recommended combination of processing conditions and steps is provided to facilitate consistent Sc contacts to WSe₂.

Introduction

Continuous engineering of contacts compatible with state-of-the-art semiconductor technology relies upon a detailed understanding of the critical relationships between processing conditions, interface chemistry and structure, and contact performance [1]. Silicides [2] and salicides [3, 4] exhibit a broad spectrum of composition-dependent contact resistances (\(R_c\)) and have long been employed as standard, low resistance contacts in traditional (Si, Ge) and compound (e.g. InGaAs) semiconductor-based CMOS technologies. Similar interface engineering has only recently been explored to improve Pd contacts to WSe₂ [5], a semiconducting member of the transition metal dichalcogenide (TMD) family of two-dimensional (2D) materials, and is a promising, versatile strategy to engineer high-performance contacts comparable with Si technology (\(R_c \approx 50 \Omega \text{cm}\)) [6].
High defect concentrations (>10^{18} \text{ cm}^{-3}) [7–9], metal-TMD and metal-ambient gas reaction products [10–12], and spurious electrostatic effects [13, 14] often manifest as strong $E_F$ pinning [15] and/or large parasitic $R_n$, which can convolute the intrinsic properties (e.g. mobility) of a TMD-based device [16, 17]. A number of strategies have been employed to reduce $R_n$ to TMDs, with varying degrees of success [18–24]. However, many are incompatible with typical backend-of-line (BEOL) process flows, and direct metallization in top contacted devices is preferred. Impressive electron contact performance in MoS$_2$ and black phosphorous (BP) devices has been demonstrated with Sc contacts [25–27]. However, Sc has yet to be explored as a contact metal in WSe$_2$-based devices. Furthermore, the highly reactive nature of Sc lends to processing difficulties and metastable device performance [26, 27]. Sc spontaneously forms scandium oxide, a high-$\kappa$ dielectric with a 5–6 eV band gap [28], even in ultra-high vacuum (UHV) conditions [10] and when a capping layer is employed to limit spurious air-exposure induced effects. Therefore, engineering high-performance Sc contacts to WSe$_2$ requires a detailed understanding of the relationship between processing conditions, interface chemistry, and Sc contact performance.

WSe$_2$ is also a promising alternative 2D switch in state-of-the-art magnetoresistive random-access memory technology due to the giant spin splitting in the valence band (456 meV) [29], moderate hole mobility [15], and low switching power. Spin-torque transfer based on the spin-valley Hall effect in WSe$_2$ has emerged as a preferred magnetic bit-writing method in analogous devices [30] but relies on a single WSe$_2$ layer, which can experience catastrophic damage during processing. Before the technology can be integrated in commercial applications, the effects of common BEOL processing conditions on the integrity of the WSe$_2$ monolayer must be quantified.

In this work, we establish relationships between processing conditions (deposition chamber ambient, post-metallization annealing temperature and ambient), interface chemistry, and band alignment in the Sc–WSe$_2$ system with in situ x-ray photoelectron spectroscopy (XPS) experiments. The number of WSe$_2$ layers consumed by reactions with Sc after fabrication, 300 °C post-metallization anneals, and as a function of time between fabrication and characterization are quantified with Raman spectroscopy, scanning transmission electron microscopy (STEM), and energy dispersive x-ray spectroscopy (EDS). Scanning tunneling microscopy/spectroscopy (STM/STS) elucidate the effects of Sc atoms on the local density of WSe$_2$ surface states, providing insight into the extracted band alignment. We demonstrate moderate $E_F$ depinning in metal-WSe$_2$ systems by deliberate growth of a scandium oxide depinning layer at the contact-WSe$_2$ interface. Finally, we evaluate the electrical performance of UHV-deposited Sc contacts to WSe$_2$ field-effect transistors (FETs) as a function of post-metallization annealing conditions. Recommendations are provided to preserve the high-performance Sc electron contacts to WSe$_2$ and to alleviate $E_F$ pinning in metal-WSe$_2$ systems, which are based upon the impressive FET performance metrics and band alignment control demonstrated here.

**Methods**

**Metallization, annealing, and in situ characterization**

(a) Depositing 1 nm Sc in UHV or high vacuum (HV); The synthetic WSe$_2$ crystals employed in this work were purchased from HQ Graphene [31]. Before performing electron beam evaporation of Sc in UHV or HV, materials (WSe$_2$, Sc metal) were prepared, the metal deposition rate was determined, and reference core level spectra were obtained identical to that employed and described in our previous work [10, 11]. Metal deposition was performed in UHV and HV using a similar base pressure of $<2 \times 10^{-9}$ and deposition pressures of $7 \times 10^{-9}$ mbar and $5 \times 10^{-6}$ mbar, respectively. The Sc depositions in HV and UHV were performed in the same chamber. However, the deposition in HV was performed by first ramping the filament current up to the deposition current under UHV conditions, then backfilling the chamber with air to a pressure of $5 \times 10^{-6}$ mbar, and subsequently opening the shutters to start the deposition. This method permits in situ XPS characterization after Sc deposition in HV unlike when the deposition is performed ex situ. A 50 nm thick Sc film was deposited on highly oriented pyrolytic graphite (HOPG) at RT in UHV and the reference Sc 2p core level spectrum was immediately obtained in situ.

(b) Step-wise Deposition and Post-Metallization Annealing: All Sc depositions, anneals, and XPS characterization were performed in a cluster tool described elsewhere [32, 33]. Sc was deposited in UHV on separate bulk WSe$_2$ crystals to estimated thicknesses of 0.1, 0.2, 1.0, 2.0, and 5.0 nm. Consecutive annealing steps were subsequently performed on each sample in UHV or forming gas (FG; 5% H$_2$, 95% N$_2$; 1 mbar) at 200 °C, 300 °C, and 400 °C for 1 h each. The stainless steel gas line connecting the FG cylinder to the annealing chamber was pumped overnight to a pressure of $<2 \times 10^{-9}$ mbar to remove any adsorbed species (i.e. H$_2$O) from within the gas line before the chamber was backfilled.
with FG and the anneal was performed. XPS was performed after exfoliation and each subsequent deposition and annealing step to characterize the surface chemistry and the secondary electron cutoff. The secondary electron cutoff was measured to track the work function. The procedures employed to fit high-resolution core level spectra and construct band diagrams are discussed in detail in the supporting information (stacks.iop.org/TDM/6/045020/mmedia).

XPS instrumentation, parameters, and data analysis
XPS characterization was performed via a monochromated Al Kα source and hemispherical analyzer (Omicron EA125) with ±0.05 eV resolution were employed for XPS. The cross sectional area of the incident x-ray beam is ~7.85 × 10⁻³ cm². A 45° takeoff angle, 8° acceptance angle, and 15 eV pass energy were employed when acquiring high-resolution spectra. The analyzer was calibrated according to ASTM E1208 [34]. Spectra were deconvolved using the curve fitting software AAAnalyzer [35].

Quantifying layer number consumption: sample fabrication and characterization
WSe₂ flakes onto a SiO₂/Si substrate (270 nm thermal SiO₂). After transferring the samples into the cluster tool, the annealing chamber was pumped to a base pressure of < 2 × 10⁻⁸ mbar and then backfilled with Ar to 1 bar before annealing at 300 °C for 1 h to remove organic tape residue. Many 1 to 5 layer (1L to 5L) flakes were identified with optical microscopy, atomic force microscopy (AFM), and Raman spectroscopy. 5 nm thick Sc films were then deposited in UHV onto the WSe₂ flakes (see supporting information for more details). Immediately after metallization, select samples were annealed in situ in UHV or FG at 300 °C for 1 h. A full coverage, 10 nm thick Si capping layer was subsequently deposited in situ using electron beam evaporation to protect the Sc–WSe₂ heterostructure from spurious air-induced reactions during ex situ Raman spectroscopy. The anneals were performed before depositing the Si cap to prevent intermixing between Si and the underlying Sc–WSe₂ heterostructure at elevated temperatures.

All Raman spectra were obtained using a laser power density of 0.49 mW μm⁻² and a 0.2 cm⁻¹ detector resolution. The Raman spectra were obtained from exfoliated WSe₂ flakes after the vacuum anneal with 1 s exposure time and ten accumulations. After metalization and annealing (where applicable), Raman spectra were obtained using a 5 s exposure time and five accumulations. These parameters were carefully tuned to prevent laser-induced damage to WSe₂ (see [5] for details regarding our carefully optimized Raman spectroscopy parameters).

Raman spectra were deconvolved with AAAnalyzer to rigorously determine Raman shifts. A combination of Gaussian and Lorentzian functions was employed in the fitting process. The Lorentzian contribution varied with the number of WSe₂ layers in the probed region and was therefore held constant for each set of spectra representing a certain number of WSe₂ layers.

STEM and EDS
(a) Lamella were milled from the Si/Sc/WSe₂/SiO₂ samples, which were initially fabricated for characterization by Raman spectroscopy, using a FIB electron microscope (FEI Nova 200 Dual Beam). High-resolution STEM was performed in an aberration corrected JEM-ARM200F instrument operated at 200 kV. Images were obtained using annular bright field and high angle annular dark field modes. EDS experiments were performed in an Aztec Energy Advanced Microanalysis System according to the procedure outlined in detail elsewhere [33]. EDS line scan data were acquired point by point (acquisition time ~0.5 s/pixel), which minimizes radiation damage in the lamella and increases the noise level in the data.

(b) A lamella from a 65 nm Pd/5 nm Sc/WSe₂ diode treated with atomic hydrogen (see below for details) was cross sectioned using an FEI Dual Beam Helios Nanolab 600i SEM/FIB microscope. Annular bright field TEM images were obtained with a JEOL 2100 operated at 200 kV.

Device fabrication
(a) The backside gate dielectric was formed by depositing a 27 nm Al₂O₃ film onto a Si wafer (p⁺⁺) at 250 °C and subsequently annealing the wafer at 400 °C in FG to reduce charge traps [36]. Al was then deposited by electron beam evaporation on the reverse side of the Si wafer as the backside contact. After exfoliating WSe₂ flakes onto the Al₂O₃, source/drain contacts were defined using photolithography and Sc/Pd (20 nm/50 nm) contacts were subsequently deposited in UHV by electron beam evaporation. The sample was then transferred ex situ to an elastomer sealed electron beam deposition tool where 100 nm Au was deposited on top of the Sc/Pd layers. Finally, a lift-off process was performed. The devices were electrically characterized in air at RT and 1 bar in a Cascade Probe Station using a Keithley 4200 Semiconductor Characterization System.
Schottky Diodes: A bulk WSe$_2$ crystal was exfoliated, resulting in a mirror-like (presumably low defect density) surface, and loaded into a UHV cluster tool [32, 33]. Select samples were treated with atomic hydrogen in situ prior to metallization. Sc/Pd contacts (65 nm Pd, see main text for Sc thickness details) were deposited through a shadow mask in UHV and 100 nm Au was subsequently deposited ex situ in an elastomer-sealed Temescal BJD-1800 electron beam evaporator [37] (base pressure $< 5 \times 10^{-6}$ mbar) to form arrays of circular contacts (diameters = 50, 100, 200 $\mu$m) across the WSe$_2$ crystal. $I$–$V$ curves were obtained by first sweeping from 0 to 2 V and then from 0 to $-2$ V (0.01 V step) to prevent hysteresis effects. Measurements were obtained after metallization and subsequent anneals in FG (1 mbar) at 200 °C, 300 °C, and 400 °C. $I$–$V$ curves were obtained from all working diodes, normalized according to the area, and directly compared to confirm the electrical performance scales with area as expected.

Atomic hydrogen treatment

An MBE Komponenten Hydrogen Atom Beam Source (Model No. HCS-40-K-2000654) with tungsten filament was operated at a filament temperature of 1500 °C and a H$_2$ (99.99999% purity) partial pressure of $5 \times 10^{-6}$ mbar. The bulk WSe$_2$ samples were maintained at a substrate temperature of 300 °C throughout the treatment, which was performed for 45 min. After the treatment and cooling to RT, the surface chemistry was characterized with XPS. Contacts were then deposited in situ.

Scanning tunneling microscopy and spectroscopy

A bulk WSe$_2$ crystal was exfoliated and loaded into a UHV cluster tool described elsewhere [33]. The STM/STS images and spectra were acquired at RT in the constant current mode using an etched tungsten tip. Imaging under positive (negative) bias probes filled (empty) surface states within a few eV of the $E_F$. The conductance (d$I$/d$V$ versus $V$) curves obtained in this work are each differentiated averages of 20 curves obtained sequentially at a single location. STM images are processed in the WSxM software.

Results and discussion

Effects of processing conditions on the Sc–WSe$_2$ interface chemistry and structure

Highly exothermic reactions between Sc, WSe$_2$, and background gases in vacuum

The metal–semiconductor interface chemistry can vary significantly with the deposition chamber base pressure and the deposition rate [10–12, 38]. According to the kinetic theory of gases, the impingement rate of background gases on the substrate during deposition in HV is sufficiently high for continuous metal oxidation on the substrate surface. In addition, the reaction products formed between highly reactive metals, such as Sc, and TMDs also undergo exothermic reactions with the background ambient, complicating the interface chemistry further.

Sc aggressively reacts with WSe$_2$ when deposited at RT regardless of the deposition chamber base pressure. Figure 1(a) shows the Se 3$d$, W 4$f$, and Sc 2$p$ core level spectra obtained from exfoliated, bulk WSe$_2$ after depositing ~1 nm Sc at RT in UHV or HV. When deposited in UHV, Sc completely reduces WSe$_2$ to form metallic W and ScSe$_x$. The presence of metallic W is confirmed by the binding energy (BE) of the asymmetrically shaped, low BE chemical state in the corresponding W 4$f$ core level spectrum (W 4$f_{7/2}$ BE = 31.30 eV), which is in close agreement with that of a metallic W reference (see Methods for details regarding the metallic W reference). The ScSe$_x$ chemical state in the Se 3$d$ core level spectrum is detected at lower BE from the WSe$_2$ chem-
ical state, which is expected considering, for example, the Pauling scale electronegativity of Sc (1.36) is much less than that of W (2.36) [39].

Our previous work investigating the interface chemistry between transition metals and TMDs have shown that early transition metals typically oxidize in situ when deposited in HV [10–12]. However, the HV deposition was performed ex situ from the post metallization XPS in [10, 11]. In the aforementioned experimental design, it is difficult to explicitly determine whether the observed oxidation occurs in situ during metallization or while transferring the sample between the elastomer-sealed deposition tool and the UHV cluster tool. In this work, the HV Sc deposition was performed in the same chamber as the UHV deposition. However, the chamber was backfilled with air to 5 × 10⁻⁶ mbar before the deposition to simulate the conditions typically found in an elastomer sealed deposition tool. This experimental design eliminates any spurious air-exposure induced changes in interface chemistry and elucidates the true chemistry in the Sc–WSe₂ system formed in HV. Sc was also deposited ex situ in an elastomer-sealed deposition chamber and subsequently characterized by XPS to compare the Sc–WSe₂ interface chemistry formed in HV with and without air exposure between deposition and XPS.

Sc reduces WSe₂ when deposited in HV with and without the air-exposure step between Sc deposition and XPS. When Sc deposition and subsequent XPS are performed in situ, the presence of substoichiometric WSe₂ and ScSeₓ are evidenced by the chemical states detected at 55.00 eV (31.73 eV) and 54.05 eV, respectively, in the Se 3d (W 4f) core level spectrum. When the HV Sc deposition and subsequent XPS are performed ex situ, the WSe₂ that presumably forms as Sc is deposited is oxidized, as evidenced by the appearance of W-Se₂Oₓ and WOₓ chemical states at higher BE from the bulk WSe₂ chemical state in the W 4f core level (figure 1(a)). However, in a typical contact structure where a thicker Sc film and inert capping metal are employed, W-ScOₓ and WOₓ are likely absent from the interface.

The additional grey peaks at low BE relative to the Se 3d and W 4f core levels correspond with the Sc 3s and Sc 3p core levels, respectively. The BE and area of these peaks were carefully calibrated according to a Sc reference film to maximize the accuracy of the fit (see supporting information for more details). When deposited in UHV, the majority of the ~1 nm Sc film reacts to form either ScSeₓ or Sc₂O₃ (figure 1(b)). In contrast, all of the Sc deposited in HV is oxidized (including or excluding air-exposure between deposition and XPS). When Sc is deposited in situ in HV, a small concentration of ScSeₓ is detected according to the low BE chemical state in the corresponding Sc 2p core level spectrum. This indicates that a small concentration of Sc reacts in situ with the underlying WSe₂ when deposited in HV. Therefore, a Sc contact deposited in HV is mostly oxidized in situ, likely implicating contact performance considering Sc₂O₃ has been employed as a high-κ dielectric [40]. Nearly complete Sc oxidation in situ in HV is reasonable considering Sc–O bond formation is highly exothermic (∆Gₓ(ScOₓ) = −630 kJ mol⁻¹) compared with the persistence of Sc–Se bonds (∆Gₓ(ScSeₓ) = −360 kJ mol⁻¹) [41]. The presence of Sc₂O₃ and Sc(OH)ₓ are corroborated by the chemical states detected in the corresponding O 1s core level (figure S3(b)).

Any reactions between adventitious carbon, which is detected on the exfoliated WSe₂ surface at ~284.4 eV (figure S3(a)), or nitrogen in the background ambient are below the limit of detection. However, when a thicker Sc film is deposited on WSe₂ in UHV, there is evidence for the formation of ScC and/or ScN, which is discussed in greater detail later.

Complete Sc oxidation at elevated temperatures in UHV

A complete understanding of the relationship between processing conditions, interface chemistry, and contact performance is critical to engineering the Sc–WSe₂ interface for high-performance electron transport. Post-metallization annealing can drive additional reactions and concomitant Eᵣ shifts depending on the temperature and ambient. Therefore, the interface chemistry and band alignment between Sc and WSe₂ were tracked in situ throughout stepwise Sc deposition in UHV and post metallization annealing (see Methods for experimental details).

Figure 2(a) displays the evolution of integrated intensities of chemical states associated with Wₓ, ScₓSeₓ, ScₓOₓ, and metallic Sc after depositing ~5.9 nm Sc on WSe₂ and subsequent UHV anneals. The integrated intensities displayed include both spin orbit split peaks in each of the Sc 2p, Se 3d, and W 4f core levels and are corrected by the appropriate atomic sensitivity factors unique to the detector employed (see supporting information). Sc reacts aggressively with WSe₂ at RT to form metallic W and ScₓSeₓ. A complete discussion regarding the evolution of chemical states in the Se 3d, W 4f, and Sc 2p core level spectra throughout Sc deposition at RT up to a total film thickness of 5.9 nm is included in the supporting information. The target Sc film thickness was 5 nm, but deviations from the calibrated deposition rate can manifest as a result of using Sc pellets as the source material instead of a solid Sc slug. In addition, calculating the Sc film thickness from core level attenuation requires the density of the attenuating film, which is difficult to estimate in this particular case considering the complex chemistry, which is discussed below.

The 200 °C UHV anneal drives Sc to react with additional WSe₂, which is evidenced by increases in the intensities of the ScₓSeₓ (50.1% of the total Sc 2p core level intensity, figure 2(c)) and metallic W chemical states in the corresponding Se 3d and W 4f core level spectra (figures 2(a) and S4(a)). However, the concentration of ScₓSeₓ decreases incrementally during
the 300 °C and 400 °C UHV anneals, which indicates Sc–Se bonds are dissociated in favor of Sc–O bonds (as predicted by thermodynamics) [41]. In addition, the concentration of metallic W decreases slightly during the 300 °C and 400 °C UHV anneals, which suggests a reaction between metallic W and Se ions liberated from ScSe, result in the reformation of W–Se bonds (figures 2(a) and S4(a)).

Figure 2(b) shows the Sc 2p core level spectrum obtained after depositing 5.9 nm Sc and after each subsequent UHV anneal. 28.7% of the 5.9 nm Sc film is converted to ScSe (Sc 2p: 400.27 eV) at RT (figures 2(b) and (c)), while the other 71.3% of the film is comprised of a mixture of metallic Sc (398.89 eV), ScO (402.16 eV), ScC (397.98 eV), and ScN (400.66 eV). ScC and ScN are near the limit of XPS detection, which is why they are difficult to resolve in figure 2(b). The presence of ScC and ScN in the same Sc–WSe2 system is validated in the discussion below.

During the 200 °C and 300 °C UHV anneals, all of the deposited metallic Sc either reacts with the underlying WSe2 to form ScSe, and metallic W or with the outer ambient to form ScO, as evidenced by the dramatic intensification of the ScSe and ScO chemical states in the corresponding Sc 2p core level spectrum (figure 2(b)). 17.9% of the ScSe formed during the 300 °C UHV anneal is converted to ScO (figure 2), which is thermodynamically favorable [41]. Extending the duration or increasing the temperature of the UHV anneal would presumably increase the relative ScO concentration within the film.

After depositing 3.2 nm Sc, a carbide chemical state is detected at ~281.6 eV in the corresponding C 1s core level is detected at 281.60 eV (figure S5), which corroborates the presence of a small concentration of ScC. The formation of Sc–C bonds is exothermic at RT ($\Delta G_f^{\text{ScC}} = -164$ kJ mol$^{-1}$) [42], which suggests ScC is present during initial deposition steps, but below the limit of XPS detection until a total of 3.2 nm Sc is deposited. The BE of the chemical state (~396.5 eV) detected after depositing > 3 nm Sc and throughout subsequent UHV anneals (figure 2(b)) is in good agreement with the ScN chemical state in the N 1s core level reported previously [43] and the ScN reference film grown in this work (figure S6). The evolution of chemical states in the C 1s core level throughout stepwise Sc deposition and subsequent UHV anneals as well as a detailed chemical analysis of the ScN reference film are discussed further in the supporting information.

Chemical states consistent with Sc$_4$O$_7$ and Sc(OH)$_x$ species are detected in the O 1s core level spectrum throughout the UHV anneals (figure S7) and are discussed in greater detail in the supporting information.

Intermetallic reduction via forming gas annealing

Annealing a TMD device in a partial pressure of H$_2$ has been shown to passivate defects and improve performance [36, 44, 45]. Therefore, we investigated the effects of FG annealing on the Sc–WSe$_2$ interface chemistry and band alignment. Prior to performing the FG anneals, stepwise Sc deposition at RT on bulk WSe$_2$ results in the same Sc film thickness, interface chemistry, and $E_F$ position (figure S8) as were detected prior to the UHV anneals (figure 2).

The 200 °C FG anneal causes the metallic W chemical state in the corresponding W 4f core level to shift +0.30 eV and the ScSe chemical states in the Se 3d and Sc 2p core levels to shift −0.35 eV and +1.26 eV, respectively, relative to the chemical states detected after Sc deposition at RT. The aforementioned BE
shifts indicate the formation of a ternary W,Sc,Se compound (figures 3(b) and S8). The anneal also dissociates Sc–Se bonds, as evidenced by the small concentration of elemental Se detected at 55.30 eV in the corresponding Se 3d core level. Sc–Se bond scission is more favorable than W–Se bond scission considering the relevant bond dissociation energies (BDE_{Sc-Se} = 385 kJ mol^{-1}, BDE_{W-Se} = 418 kJ mol^{-1}) [46, 47]. After the 400 °C FG anneal, the W,Sc,Se chemical states in the W 4f (Se 3d) core level shifts +0.42 eV (−0.34 eV) relative to that detected after the 200 °C FG anneal, which is consistent with a decreased concentration of Sc within the W,Sc,Se compound and a corresponding increased oxidation state of the associated W^{VII} component (figure S8). Thermodynamics indicates an additional anneal performed either for a longer period or at a higher temperature could completely dissociate Sc from W,Sc,Se, potentially impacting contact resistance further. The gas line connecting the pressurized FG cylinder with the UHV annealing chamber was opened to the chamber overnight to remove adsorbed contaminants from the gas line before performing the FG anneals. However, the specific (presumably negligible) concentration of oxygen-based impurities within the FG and the associated effects on the scandium oxide concentration in the Sc film is not known in detail. Employing a molecular sieve in the gas line between the FG and the annealing chamber could cause additional variations in the chemistry and performance of the Sc contact to WSe_2.

The 200 °C FG anneal converts nearly all of the deposited Sc into either W,Sc,Se or Sc,O_y (figure 3(b)). The intensity of the chemical state in the Sc 2p core level corresponding with the Sc–Se bond decreases by 64.1% during the 200 °C FG anneal. The Sc–Se contribution to the total intensity of the Sc 2p core level spectrum decreases to 18% and 5% after the 300 °C and 400 °C FG anneals, respectively (figures 3(b) and (c)), which contrasts the relatively stable intermetallic concentration throughout the UHV anneals. The Sc(N and Sc(C) chemical states detected after the 200 °C FG anneal fall below the limit of XPS detection after the 400 °C FG anneal. This suggests the partial pressure of H_2 in the FG ambient dissociates Sc–C and Sc–N bonds, which is an energetically favorable process (the energy liberated when a H–H bond is broken is greater than the BDE_{Sc-C} and BDE_{Sc-N}) [46].

This work indicates a Sc contact to WSe_2 will completely oxidize when the post-metallization anneal is performed in FG, provided the anneal is performed at a high enough temperature or for a long enough time. Complete Sc–Se, Sc–C, and Sc–N bond dissociation in the presence of oxygen-containing species (i.e. the background gases in a vacuum chamber) are energetically favorable considering either the thermodynamics (see earlier sections for relevant ΔG^\circ) or the kinetics (BDE_{Sc-O} > BDE_{Sc-N} > BDE_{Sc-C} > BDE_{Se-Se} > BDE_{Se-Sc}) [46] of the system. The effects of the increased Sc,O_y concentration on the band alignment and performance of the Sc contact to WSe_2 will be discussed in detail later.

Multiple WSe_2 layers consumed by thermally exacerbated reactions with Sc

The experiments discussed above were performed on bulk WSe_2, which provides an appropriate platform to characterize effects of certain post metallization anneals on the metal-TMD interface chemistry and band alignment. However, FETs are typically fabricated with single and few layer TMDs. In addition, edge contacts exhibit superior performance compared to top contact analogs according to DFT calculations [48, 49] and experimental demonstrations [21, 49]. The true structure of the contact (edge versus top) will be affected by reactions at the metal-TMD interface. In addition, the broken inversion symmetry in WSe_2 films with D_{3h} symmetry is critical to the unique giant spin–orbit splitting in the valence band in the absence

Figure 3. (a) Integrated intensities of the W,Sc,Se intermetallic and Sc,O_y chemical states in the Se 3d, W 4f, and Sc 2p core level spectra. (b) Sc 2p core level spectra obtained from bulk WSe_2 after Sc deposition at RT and subsequent in situ FG anneals. (c) Percentage of the deposited Sc film converted to ScSe intermetallic after deposition at RT and subsequent FG anneals showing aggressive reactions at RT and increasing Sc–Se dissociation with increasing FG anneal temperature.
of an out-of-plane electric field [29]. It is therefore of interest to quantify the number of WS\textsubscript{2} layers affected by reactions with Sc.

Variations in structural and electronic properties of single and few layer TMDs due to interactions with a deposited metal have been investigated previously [30–32]. For example, depositing an incomplete coverage metal film on MoS\textsubscript{2} resulted in a complex vibrational response due to metal-induced spatially varying strain across the TMD [51]. In this work, Raman spectroscopy is employed to quantify the number of layers consumed by reactions with Sc via a characteristic, layer number-dependent vibrational mode exhibited by WS\textsubscript{2}. Special care was taken to ensure a full coverage Sc film was deposited (figure S2). In addition, a 10 nm thick Si capping layer was deposited in situ after the Sc deposition and the subsequent anneal (where applicable) to prevent any spurious, air-exposure induced changes in the Raman spectra obtained ex situ (procedure described in detail elsewhere) [5]. Therefore, any structural changes manifesting in the Raman spectra are attributed to Sc–WS\textsubscript{2} reactions. It is important to note here the Raman measurements were obtained within 24 h of Sc/Si deposition and the subsequent anneal.

The first (1st) order in-plane ($E_{1g}$) and out-of-plane ($A_{1g}$) vibrational modes of WS\textsubscript{2} are degenerate and do not exhibit any discernible characteristic shifts or changes in intensity with layer number [53]. We recently demonstrated the layer number dependent Raman shifts exhibited by the second (2nd) order longitudinal acoustic mode at the M point in the Brillouin zone [$2LA(M)$] from single layer to five layer WS\textsubscript{2} [2.5 cm\textsuperscript{-1}, 0.5 cm\textsuperscript{-1}, 0.5 cm\textsuperscript{-1}, and 0.3 cm\textsuperscript{-1}] red shifts increasing from one layer (1L, 261.3 cm\textsuperscript{-1}) to 2L, 2L to 3L, 3L to 4L, and 4L to 5L WS\textsubscript{2} [5]. Therefore, the number of WS\textsubscript{2} layers remaining after the Sc deposition and subsequent anneal can be accurately determined by tracking the Raman shift of the 2LA(M) mode. A $\lambda = 532$ nm laser is employed here to access the 2LA(M) mode via resonant excitation conditions [54]. The laser power density (0.49 mA $\mu$m$^{-2}$), number of sweeps (5), and exposure time per sweep (5 s) employed in this work were carefully selected according to control experiments performed previously [5] to prevent laser-induced WS\textsubscript{2} damage. Therefore, spectral changes were confidently interpreted as indicators of chemical interactions between Sc and WS\textsubscript{2} rather than laser-induced intermixing.

Figures 4(a)–(c) display the Raman spectra obtained from exfoliated 1L, 2L, and 3L WS\textsubscript{2} flakes before and after depositing 5 nm Sc at RT and subsequent 300 °C UHV or FG anneals. Depositing 5 nm Sc at RT completely quenches the 1st and 2nd order modes of the 1L WS\textsubscript{2} flake, which indicates 1L WS\textsubscript{2} is consumed by reactions with Sc at RT (figure 4(a)). The 1.0 and 1.1 cm\textsuperscript{-1} red shifts and slight asymmetric broadening exhibited by the 2LA(M) mode detected from the 2L and 3L WS\textsubscript{2} flakes, respectively, after depositing 5 nm Sc at RT suggest the interfacing reaction products cause stiffening (softening) of the $A_{1g}$ ($E_{1g}$) mode of the underlying WS\textsubscript{2} [51, 55].

After the 300 °C UHV anneal, the 1st and 2nd order vibrational modes of the 1L and 2L WS\textsubscript{2} flakes are completely quenched, while the 1st order modes exhibited by the 3L flake are near the limit of detection (figure 4(b)). This suggests at least two and possibly three WS\textsubscript{2} layers are consumed by reactions catalyzed during the 300 °C UHV anneal. To more explicitly quantify the number of layers consumed during the anneal, the vibrational modes of a 5L WS\textsubscript{2} flake were also probed throughout the experiment (figure 4(b)). After the anneal, the corresponding 2LA(M) mode exhibits a 1.2 cm\textsuperscript{-1} blue shift, which is consistent with a transition from 5L to 2L WS\textsubscript{2} [5]. The spectrum obtained from exfoliated 2L WS\textsubscript{2} (dotted line) is normalized to the 5L WS\textsubscript{2} spectra to clearly show the similarity between the spectrum of pristine 2L WS\textsubscript{2} and 5L WS\textsubscript{2} after Sc deposition and subsequent 300 °C UHV anneal. Therefore, we confidently conclude that three WS\textsubscript{2} layers are consumed by reactions with Sc during the 300 °C UHV anneal.

Interestingly, 1L WS\textsubscript{2} exhibits 1st and 2nd order modes above the limit of detection after the 300 °C FG anneal despite significant reactions detected by XPS in this work. In addition, asymmetric broadening towards lower wavenumber is detected in the 1st order modes of all WS\textsubscript{2} flakes after the FG anneal. This behavior indicates Se$^{2-}$ (liberated by Sc–Se bond scission) reacts with metallic W to form defective WS\textsubscript{2} clusters, which is consistent with the Raman spectrum obtained from a WS\textsubscript{2} film lacking long range order [55] and also the XPS results displayed in figure 3.

Roughly nine months after characterizing the three different 10 nm Si/5 nm Sc/3L WS\textsubscript{2}/270 nm SiO\textsubscript{2}/Si samples with Raman spectroscopy (figures 4(a)–(c)), lamella were milled and imaged by STEM. The corresponding STEM images are shown in figures 4(d)–(f). After Sc deposition at RT (figure 4(d)), nanocrystalline grains with lattice spacing of 0.21 nm are observed in the Sc film, which is consistent with the Sc{111} family of planes [56]. A significant concentration of oxygen is detected by EDS throughout the Sc film (figure S9). However, the lattice spacing of the grains observed in figure 4(d) is not consistent with that of Sc$_2$O$_3$ (0.31 nm) [57], which indicates the oxygen is dispersed in amorphous regions of the film between grains. Sc and W diffusion into the Sc film is observed in the corresponding EDS (figure 4(d)). The three ~0.7 nm thick stripes of dark contrast coincide with high-Z W atoms and therefore indicate the presence of three WS\textsubscript{2} layers (figure 4(d)). The top most WS\textsubscript{2} layer in figure 4(d) appears to have retained its planar structure in some regions, but the contrast is slightly different from the underlying two layers indicating some disruption of the top layer due to intermixing in general agreement with the corresponding Raman spectra in figure 4(a).
When a lamella is milled from an analogous Sc–WSe₂ sample fabricated at RT in UHV (see supporting information for details) within two weeks of fabrication and imaged immediately, a 2.0–2.5 nm thick amorphous region is observed between Sc and WSe₂ in the corresponding TEM images (figure S10). Therefore, the WSe₂ involved in reactions with Sc undergoes restructuring over time as Sc oxidizes, leading to the physical differences observed by TEM in this work depending on the time between fabrication and imaging.

The thickness of the dark contrast region observed after the 300 °C UHV anneal (~2.0 nm, figure 4(e)) is consistent with that expected of pristine 3L WSe₂. However, individual layers are no longer distinguishable, which suggests significant intermixing occurs between WSe₂ and Sc and corroborates three WSe₂ layers are consumed by reactions during the 300 °C UHV anneal. An amorphous region is observed between the disordered WSe₂ and the polycrystalline Sc film, which likely corresponds with the ScSeₓ intermetallic detected by XPS after the same anneal. EDS indicates Se and W diffuse ~3 nm up into the Sc film and down into the SiO₂ substrate. The 0.30 nm lattice spacing exhibited by the nanocrystallites in the Sc film after the anneal (figure 4(e)) indicates the dramatic increase in scandium oxide detected by XPS after the same anneal (figure 2) corresponds with the formation of a polycrystalline Sc₂O₃ film.

After the 300 °C FG anneal, local atomic structure and interlayer van der Waals gaps are resolvable in the corresponding STEM image (figure 4(f)). EDS indicates Se and W diffuse 2–3 nm into the Sc film, while Sc, Se, and W diffuse up to 5 nm into the underlying SiO₂ (dark regions below the WSe₂). Metal diffusion into the underlying dielectric in back-gated devices could impact the device performance (e.g. decreased gate modulation, increased off current). The WSe₂ film is comprised of disordered regions with limited atomic order adjacent regions of 3L WSe₂ with clearly resolvable van der Waals gaps between each layer. These observations corroborate our hypothesis.

Figure 4. Raman spectra displaying peaks corresponding with the \( E_{12g}, A_{1g} \), and 2LA(M) vibrational modes of WSe₂ obtained from 1L, 2L, 3L, and 5L (where applicable) flakes prior to and following (a) 5 nm Sc deposition under UHV conditions and subsequent post metallization anneal at 300 °C in (b) UHV or (c) forming gas. All spectra are normalized to the 2LA(M) peak unless vibrational modes are below the limit of detection. (d)–(f) Cross section STEM images and associated EDS spectra obtained from the same 3L WSe₂ flakes investigated by Raman spectroscopy.
based on the corresponding XPS (figure 3) and Raman results; W–Se bonds re-form via Sc–Se bond dissociation throughout the FG anneals.

In a device where the contact metal consumes at least one TMD layer, the resulting intermetallic likely remains in intimate lateral contact with the adjacent channel. Therefore, any TMD layers in the contact region consumed by reactions at the metal-TMD interface should be considered pseudo-edge contacts, which may exhibit superior performance to the top contact analog. In addition, understanding changes in the band structure of few layer TMDs associated with interface reaction-induced layer number thinning is critical to engineering superior performance in a wide variety of TMD devices. For instance, maintaining an odd number of layers in the channel of a WSe$_2$ spin valve is critical to device operation.

**Band alignment and electrical performance of the Sc contact to WSe$_2$**

The $E_F$ shifts towards the conduction (valence) band edge will be referred to in the following discussion as positive (negative). Figure 5(a) displays the absolute BEs of the WSe$_2$ chemical state in the Se 3$d$ core level spectra throughout stepwise Sc deposition and post-metallization annealing in either UHV or FG conditions. The BE shifts throughout RT Sc deposition are similar for both samples, which is why the data points obtained from the ‘FG anneal’ sample prior to the anneals are displayed in (a). Band alignments after the samples are annealed in (b) UHV and (c) FG at 400°C, which are derived from XPS measurements. The error of ±0.1 eV is associated with all Fermi level positions depicted. (d) $I_{DS}-V_{BG}$ characteristics (solid lines) measured from back gated few layer WSe$_2$ FETs with Sc contacts after fabrication and subsequent annealing in UHV or FG. The symbols in (d) correspond with fits of the $I_{DS}-V_{BG}$ curves in (d). Band diagrams in (b) and (c) are reflected in (e). The ambipolar FET characteristics obtained after FG and UHV anneals corroborate the mid-gap band alignment indicated by XPS.

After exfoliation, the $E_F$ is detected 0.95 ± 0.15 eV from the WSe$_2$ valence band edge according to the initial valence band offset and secondary electron cutoff (0.95 ± 0.07 eV, 4.36 ± 0.08 eV, respectively; figure S11). The bulk WSe$_2$ chemical states initially shift to lower BE during the first two Sc depositions. In contrast, the $E_F$ shifts towards higher BE beyond an effective Sc film thickness of 5 Å. The WSe$_2$ chemical states formed during initial Sc depositions likely shift the $E_F$ towards the valence band, while the metallic Sc that accumulates in latter depositions shift the $E_F$ towards the conduction band. Depositing 5.7 nm Sc at RT in UHV shifts the WSe$_2$ chemical states −0.09 eV (figure 5(a)) from the BEs detected after exfoliation, which corresponds with the formation of a 0.44 ± 0.15 eV electron Schottky barrier. The appreciable Schottky barrier detected here is far from the expected Ohmic electron band alignment expected between Sc and WSe$_2$. The separation between the WSe$_2$ chemical states in the Se 3$d_{5/2}$ and W 4$f_{7/2}$ core level spectra were held constant throughout Sc deposition and post metallization annealing.

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Therefore, the binding energies of the WSe$_2$ chemical state in the W 4$f_{7/2}$ core level spectra are omitted from figure 5(a).
WSe$_2$ considering the low metal work function (3.5 eV) [58]. It is possible the metallic W, formed as a product of the Sc–WSe$_2$ reaction, dominates the band alignment in an unannealed Sc contact to WSe$_2$ (polycrystalline tungsten work function $\approx$4.5 eV) [58].

As the UHV anneal temperature increases, the $E_F$ shift towards lower BE increases in magnitude, exhibiting a total $-$0.26 eV shift after the 400 °C UHV anneal. The $E_F$ shift corresponds with an increased electron SBH to $0.70 \pm 0.15$ eV (figure 5(b)).

The band alignment between Sc and a separate bulk WSe$_2$ crystal after exfoliation, Sc deposition, and subsequent FG anneals (figures 5(a) and (c)) is similar with that detected in the ‘UHV anneal’ sample. A more detailed discussion of the band alignment between Sc and bulk WSe$_2$ throughout annealing in FG is provided in the supporting information.

$E_F$ pinning occurs due to gap states in the semiconductor that can be generated by defects, interface reactions, or a deliberately placed re-pinning layer [59]. An appreciable concentration of oxygen deficient Sc$_2$O$_3$ in the Sc film could generate gap states in the underlying WSe$_2$, producing the near-midgap alignment detected after the 400 °C anneals. When annealing is performed in UHV, the chemistry, and therefore the electrostatics of the junction, are also convoluted by Sc$_x$Se$_{1-x}$. The Ohmic hole band alignment reported recently between Pd and WSe$_2$ after a 400 °C FG anneal was facilitated by passivating defects at the interface with atomic hydrogen in the FG ambient [5]. However, the beneficial effects of hydrogen-induced defect passivation on the Sc contact to WSe$_2$ are either negligible compared with the interface chemistry effects or require hydrogen radicals, which are not readily catalyzed from H$_2$ by Sc as they are in Pd.

To corroborate the XPS-derived near-midgap band alignment with the electrical performance of Sc contacts to WSe$_2$, back-gated, few-layer WSe$_2$ FETs were fabricated on an Al$_2$O$_3$/Si substrate with 20 nm Sc/50 nm Pd/100 nm Au contacts. It is difficult to accurately extract SBHs from ultra-thin body transistors in which the depletion width is defined by the thickness of the region under the contacts, such as in few-layer TMD FETs. Therefore, an analytical Schottky barrier model based on Landauer transport theory [14] was employed here to extract electron and hole SBHs from the measured $I_{DS}$–$V_{BG}$ characteristics of our few layer WSe$_2$ FETs with Sc contacts. The model, which has been employed in recent works [14, 15, 26, 27], accounts for thermionic emission and the appreciable tunneling contribution to the total current due to the ultra-thin body of the devices. After fabrication, the Sc–WSe$_2$ FETs exhibit impressive $I_{ON}/I_{OFF}$ ratios in the order of $10^6$, subthreshold slope (SS) of 109 mV dec$^{-1}$ (among the best reported to date) [16], and electron SBHs of 0.4–0.45 eV (figure 5(d)), which are in good agreement with the corresponding XPS-derived 0.36–0.44 eV electron SBH (figure 5(e)). Analogous WSe$_2$ FETs with Pd contacts exhibit a larger 0.73 eV electron SBH, which is likely attributed to photoresist residue-induced $E_F$ pinning [5, 24]. Sc mitigates $E_F$ pinning effects induced by photoresist residue in as-fabricated devices by reacting with the WSe$_2$ at RT and ‘cleaning’ resist residue from the interface.

Controling the contact polarity by inserting a Sc$_2$O$_3$ interlayer

$E_F$ depinning at the contacts in devices based on 2D materials has been achieved by inserting a tunnel barrier [15, 60] or ‘re-pinning’ layer [59] between the contact and the channel. Sc$_2$O$_3$ has only recently been explored, albeit unintentionally, as an interlayer to control contact performance [27]. In this work, WSe$_2$ was treated with atomic hydrogen to etch the surface, which promotes surface oxidation in situ in UHV (figure S12, see supporting information for details). Pd/Sc contacts were then deposited in situ, in UHV, at RT, and through a shadow mask to reactively form a Sc$_x$O$_y$ interlayer between the metal and the WSe$_2$. Sc scavenges oxygen from WO$_3$ during the UHV deposition, forming an interfacial Sc$_x$O$_y$ layer. A 3 nm Sc film was deposited on one sample to form
an interlayer entirely comprised of intermetallics and Sc\textsubscript{2}O\textsubscript{3}, and therefore facilitate hole dominant conduction via the Pd layer. 15 nm Sc was deposited on another atomic hydrogen-treated sample to form a substantial metallic Sc layer between the Sc\textsubscript{2}O\textsubscript{3} interlayer and the Pd capping layer, which facilitates electron dominant conduction.

The forward bias currents measured from the 65 nm Pd/5 nm Sc/WSe\textsubscript{2} diodes were transformed according to equation (1) \[62\]

$$I = \frac{1}{1 - e^{-\frac{qV}{kT}}}$$

where \(I\), \(q\), \(V\), \(k\), and \(T\) represent the measured current, electron charge, applied bias, Boltzmann’s constant, and temperature, and the linear region in each curve between 0.15 V and 0.35 V was fitted using linear regression. The Richardson constant (33.9 A cm\textsuperscript{-2}K\textsuperscript{-2}, \(m^*_{e} = 0.33\) \[63\]), SBH (0.85 eV), and an ideality factor (1.00, averaged from 5 diodes) are calculated from the slope and intercept of the best fit line to the \(I-V\) curve obtained at 110 \(^\circ\)C (figures S13(a) and (b), see supporting information for details) \[62\]. An ideality factor of 1.00 indicates the barrier height is homogeneous across the interface. This suggests the atomic hydrogen treatment coupled with the Sc\textsubscript{2}O\textsubscript{3} interlayer eliminate defects at the interface unlike the highly inhomogeneous barrier formed in the inert Pd–WSe\textsubscript{2} interface at RT. However, the \(E_{\text{F}}\) resides much closer to midgap when the Sc\textsubscript{2}O\textsubscript{3} interlayer is included between Pd and WSe\textsubscript{2} compared with the Ohmic hole band alignment exhibited by Pd contacts to WSe\textsubscript{2} at RT \[5\]. Some metallic Sc is likely present within the interlayer and contributes to the near-midgap alignment measured here.

The transformed, annealing-temperature-dependent \(I-V\) curves obtained from the 65 nm Pd/15 nm Sc/ WSe\textsubscript{2} Schottky diodes yield ideality factors > 3 in all cases, which indicates the corresponding SBH will be unreliable. However, an Arrhenius plot of the reverse bias \(I-V\) characteristics of the 65 nm Pd/15 nm Sc/WSe\textsubscript{2} Schottky diodes yields a linear pattern, from which a 0.34 eV electron SBH is extracted. The 0.34 eV electron SBH obtained with the 15 nm Sc layer deposited onto the atomic hydrogen-treated WSe\textsubscript{2} is 0.51 eV smaller than the diodes where only 3 nm Sc was deposited (figures 6 and S13(c)) and \(-0.1\) eV less than Sc–WSe\textsubscript{2} FETs after metallization at RT. The contact structures, fabrication and processing details, and employing an optimized, oxygen deficient Sc\textsubscript{2}O\textsubscript{3} interlayer will minimize reactions with WSe\textsubscript{2} and also provide greater \(E_{\text{F}}\) control when other contact metals are employed in conjunction with the interlayer.

**Local density of states around Sc atoms/ clusters on WSe\textsubscript{2}: STM and STS**

The Sc–WSe\textsubscript{2} diodes and FETs discussed above exhibit appreciable 0.4–0.45 eV electron SBHs at RT, which is unexpected considering the work function of Sc (~3.5 eV) is smaller than the electron affinity of WSe\textsubscript{2} (~4.1 eV) \[9, 61\]. Scanning tunneling microscopy (STM) and spectroscopy (STS) elucidate the surface topography and local density of states (LDOS). Therefore, STM and STS can provide insight into changes in the electronic structure of the WSe\textsubscript{2} surface induced by Sc in the earliest stages of metallization (i.e. after depositing a <3 Å Sc film). The exfoliated WSe\textsubscript{2} surface before metallization is atomically flat with a random distribution of atomic scale defects across the surface, similar to previously published STM images from bulk WSe\textsubscript{2} \[9\]. The upper and lower STM images in figure 7(a) were obtained \textit{in situ} under positive and negative tip bias, respectively, from exfoliated, bulk WSe\textsubscript{2} after depositing \(-1\) Å Sc in UHV at RT. Sc atoms/ clusters appear as small bright spots surrounded by large dark regions under forward bias, while they appear as large bright patches and ~2 nm diameter rings under reverse bias. The dark contrast exhibited by Sc regions in positive bias indicate an associated concentration (i.e. minimizing intermetallic-induced gap states). The contact structure, fabrication and processing details, and employing an optimized, oxygen deficient Sc\textsubscript{2}O\textsubscript{3} interlayer will minimize reactions with WSe\textsubscript{2} and also provide greater \(E_{\text{F}}\) control when other contact metals are employed in conjunction with the interlayer.
high resistance tunneling barrier and therefore fewer occupied LDOS. Sc oxidation occurs in UHV and therefore could contribute to the suppressed LDOS around Sc when ‘filled’ states are probed.

Five STS spectra were obtained in an array spanning a Sc cluster and the surrounding WSe₂ (dots in figure 7(a), lower image). The bare WSe₂ exhibits a band gap of 1.32 ± 0.05 eV, which is consistent with previous STM studies of bulk WSe₂ (figure 7(b)) [9]. The E_F is detected 0.33 ± 0.05 eV from the valence band edge, which is consistent with the XPS-derived band alignment of exfoliated WSe₂ in this work. As the tip approaches the Sc cluster, the reverse bias conductance and density of gap states below the E_F increase and reach a maximum when the tip is positioned directly over the Sc cluster. Each spectrum is a differentiated average of 20 sweeps over the same spot. The spectra are manually offset to more clearly show the features in each spectrum. (c) High resolution STM images obtained from the same 1 Å Sc–WSe₂ sample showing atomic resolution in a region of bare WSe₂, rings of bright contrast reflecting charge redistribution around single Sc atoms (some indicated by green arrows), and larger Sc clusters.

The ring of bright contrast observed in reverse bias images manifests as a result of charge transfer from Sc to the surrounding Se atoms, similar to the charge redistribution associated with mirror twin boundaries in TMDs. Therefore, the formation of Sc–Se bonds and the associated enhanced density of states below the WSe₂ E_F contribute significantly to the unexpectedly high electron Schottky barrier formed between the Sc contact and WSe₂.

Discussion

The physical characterization, E_F shifts according to XPS, and electrical characteristics of Schottky diodes and FETs indicate the WSe₂ FETs exhibit the lowest electron SBH, the highest I_ON/I_OFF ratio (~10⁶), and the steepest SS (10⁹ mV dec⁻¹) when the intermetallic concentration at the Sc–WSe₂ interface is minimized and the concentration of metallic Sc within the contact is maximized. In this work, the lowest electron SBH between Sc and WSe₂ (0.34 eV) is achieved by oxidizing the WSe₂ surface prior to Sc deposition, which both limits the Sc–WSe₂ reactions and decreases oxygen concentration in the

<table>
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<th>Sc Thickness</th>
<th>Pd Thickness</th>
<th>Anneal Temperature</th>
<th>Anneal Ambient</th>
<th>Atomic Hydrogen Treatment</th>
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<td>N/A</td>
<td>45 min</td>
<td>0.34 eV</td>
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Figure 7. (a) STM images obtained from WSe₂ after a ~1 Å Sc film was deposited in UHV with the upper (lower) image obtained at 1.5 V and 0.5 A (~1.8 V and 0.8 A). (b) STS conductance spectra obtained from points spanning a ~3 nm diameter Sc cluster and the surrounding WSe₂ showing the significant increase in density of states below the WSe₂ E_F when the tip probes directly over the Sc cluster. Each spectrum is a differentiated average of 20 sweeps over the same spot. The spectra are manually offset to more clearly show the features in each spectrum. (c) High resolution STM images obtained from the same 1 Å Sc–WSe₂ sample showing atomic resolution in a region of bare WSe₂, rings of bright contrast reflecting charge redistribution around single Sc atoms (some indicated by green arrows), and larger Sc clusters.
Sc layer compared to the reactively formed \( \text{Sc}_2\text{O}_3 \) interlayer employed in this work, likely resulting in the highest performance Sc contact to WSe\(_2\). Sc immediately consumes one WSe\(_2\) layer at RT and at least three layers during the post-metallization anneals. This work establishes relationships between the Sc–WSe\(_2\) interface chemistry, structure, and band alignment associated with specific pre- and post-metallization processing steps, which are integral to engineering consistent, high-performance Sc contacts to any TMD. Critically, we demonstrate high performance n-type WSe\(_2\) FETs with Sc contacts and establish processing conditions (both to employ and to avoid) for consistent, high-performance n-type Sc contacts.

Conclusions

This work demonstrates high-performance n-type WSe\(_2\) FETs with impressive \( I_{\text{ON}}/I_{\text{OFF}} \) (10\(^6\)) and SS (109 mV dec\(^{-1}\)), which is achieved without any post-metallization processing and by depositing the Sc contacts in UHV. Mildly oxidizing the WSe\(_2\) surface at the contact regions before metallization reduces the electron SBH formed between Sc and WSe\(_2\) at RT by 0.10 eV as a direct result of the minimized concentration of scandium selenide at the interface. The largest electron SBHs of 0.70 eV and 0.87 eV (400 °C UHV and FG, respectively) worst SS (~1 V dec\(^{-1}\)), and lowest \( I_{\text{ON}}/I_{\text{OFF}} \) ratios (~10\(^3\)) are measured when post-metallization anneals are employed, which contrasts the highest performance Pd contacts to WSe\(_2\) after the same anneal. XPS indicates the anneals increase the electron SBH by 0.4–0.5 eV and completely oxidize the Sc contact, which cause the aforementioned degraded device performance. STM and STS explicitly relate the unexpectedly large electron SBH observed between Sc and WSe\(_2\) throughout this work with an enhanced LDOS below the WSe\(_2\) \( E_F \) in the presence of Sc–Se bonds, which necessitates processing steps that minimize the Sc–WSe\(_2\) reaction. The significant reactions between Sc and WSe\(_2\) are corroborated by Raman spectroscopy and STEM, which indicate IL WSe\(_2\) is consumed at RT and at least three WSe\(_2\) layers are consumed during the anneals. The processing condition-dependent number of WSe\(_2\) layers consumed by Sc is a critical benchmark for future device architectures based on WSe\(_2\), especially those relying on the giant spin–Hall effect that occurs in WSe\(_2\) with \( D_{3h} \) symmetry. This work shows a detailed understanding of the relationships between processing conditions, interface chemistry, and contact performance can be leveraged in any metal–TMD system to consistently achieve the highest device performance.

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Supporting information

C 1s and O 1s core levels obtained from WSe\(_2\) after Sc deposition in UHV and HV; calibrating the Sc 3s and Sc 3p core level BE and intensity; calculating stoichiometry based on XPS; chemical state evolution throughout stepwise Sc deposition and post metallization annealing in UHV; evidence of Sc–C, Sc–N, and Sc–O bonding; chemical state evolution throughout Sc deposition and post metallization annealing in FG; determining the critical Sc and Si thickness for full coverage films; EDS including the oxygen spectra; TEM image of the Pd–Sc–WSe\(_2\) structure two weeks after fabrication; SBH extraction from Schottky diode \( I–V \) characteristics; constructing band diagrams from XPS measurements; oxidized WSe\(_2\) after atomic hydrogen treatment; valence band edge and secondary electron cutoff from exfoliated WSe\(_2\); AFM images of exfoliated single and few layer WSe\(_2\) flakes.

Conflict of interest

The authors declare no competing financial interest.

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