Helmholtz-Zentrum Dresden-Rossendorf (HZDR)



Formation of In-Plane Semiconductor-Metal Contacts in 2D Platinum Telluride by Converting PtTe₂ to Pt₂Te₂

Lasek, K.; Li, J.; Ghorbani Asl, M.; Khatun, S.; Alanwoko, O.; Pathirage, V.; Krasheninnikov, A.; Batzill, M.;

Originally published:

November 2022

Nano Letters 22(2022)23, 9571-9577

DOI: https://doi.org/10.1021/acs.nanolett.2c03715

Perma-Link to Publication Repository of HZDR:

https://www.hzdr.de/publications/Publ-35498

Release of the secondary publication on the basis of the German Copyright Law § 38 Section 4.

Formation of In-Plane Semiconductor-Metal Contacts in 2D Platinum Telluride by Converting $PtTe_2$ to Pt_2Te_2

Kinga Lasek,^{1,§} Jingfeng Li,^{1,§} Mahdi Ghorbani-Asl,² Salma Khatun,¹ Onyedikachi Alanwoko,¹ Vimukthi Pathirage,¹ Arkady V. Krasheninnikov,^{2,3} Matthias Batzill^{1,#}

¹Department of Physics, University of South Florida, Tampa, FL 33620, USA

² Helmholtz-Zentrum Dresden-Rossendorf, Institute of Ion Beam Physics and Materials Research, 01328 Dresden, Germany

³Department of Applied Physics, Aalto University, P.O. Box 11100, 00076 Aalto, Finland

[#]Corresponding author: mbatzill@usf.edu

Abstract:

Monolayers of platinum tellurides are particularly interesting 2D materials because they exhibit phases with different stoichiometries and electronic properties. Specifically, PtTe₂ is a narrow gap semiconductor while Pt₂Te₂ is a metal. Here we show that the former can be transformed into the latter by reaction with vapor-deposited Pt atoms. Owing to low surface diffusion barriers of Pt ad-atoms, the transformation occurs by nucleating the Pt₂Te₂ phase within the PtTe₂ islands, so that a metal-semiconductor lateral junction is formed. Using scanning tunneling microscopy/spectroscopy, the electronic structure of this lateral junction is studied. A flat band structure is found with the Fermi-level of the metal aligning with the Fermi-level of the intrinsically p-doped PtTe₂ suggesting low contact resistance. This flat band is achieved by an interface dipole that accommodates the ~0.2 eV shift in the work functions of the two materials. First-principles calculations indicate that the origin of the interface dipole is the atomic scale charge redistributions at the heterojunction. The demonstrated compositional phase transformation of a 2D semiconductor into a 2D metal is a promising approach for making in-plane metal contacts that are required for efficient charge injection and is of particular interest for semiconductors with large spin-orbit coupling, like PtTe₂.

TOC-graphic:



§ Contributed equally

Phase transformations in 2D materials and development of heterostructures based on the compounds with the same chemical composition have recently received considerable attention due to the possibility of post-synthesis modifications of existing structures. In particular, switching between semiconducting and metallic phases of 2D sheets of transition metal dichalcogenides (TMDs) has been accomplished by structural phase changes from the 2H-semiconducting to 1T'-metallic phase in group VIB TMDs.¹ Such phase transformations can be achieved by increasing vacancy concentration due to thermal treatment,² electron irradiation,³ or charge doping by alkali metal atom adsorption.^{4,5,6,7,8} Since in most cases a structural switch occurs from the lower energy H-phase to a higher energy T'-phase,^{9,10} these transformations are reversible, and the system can return back to the more stable H-phase. In contrast, compositional phase changes in 2D materials^{11,12,13} are not spontaneously reversible and thus the produced phases are stable. However, there are few examples of compositional control of contacts and usually these are mixtures of 2D and 3D crystal structures.¹⁴

PtTe₂ is an intriguing material with strongly layer dependent electronic properties. Only the monolayer is semiconducting while multi-layers and bulk materials are metallic.¹⁵ Moreover, the platinum-tellurium system exhibits various layered compositional phases, including PtTe₂ and Pt₂Te₂.^{16,17,18,19} While monolayer PtTe₂ is semiconducting with a band gap of ~0.5 eV,¹⁵ Pt₂Te₂ is metallic. Both phases possess a hexagonal structure with very closely matched lattice constants, so that a heterojunction between the two phases can be constructed with virtually no interface strain and defects. Here we demonstrate a low temperature process that enables the transformation of PtTe₂ into Pt₂Te₂ by incorporation of vapor deposited Pt atoms into PtTe₂ monolayer. The partially converted monolayer flakes exhibit PtTe₂/Pt₂Te₂ heterojunctions, which enables the spectroscopic analysis of the semiconductor/metal interface in a planar system. In contrast to previous modifications of TMDs with excess transition metal atoms, ^{20,21,22,23} we consider here stable extended 2D phases, *i.e.*, the transformation from one known 2D phase to a 2D phase with another stoichiometry of the same elements. The transformation of PtTe₂ to Pt₂Te₂ may be referred to as an introduction of an additional layer of Pt as schematically illustrated in Fig. 1, without the need for rearrangement of the Pt-atoms in PtTe₂. This close structural relationship between PtTe₂ and Pt₂Te₂ is expected to facilitate a compositional transformation by incorporation of excess Pt and the formation of well-defined heterojunctions.



PtTe₂ O Te O Pt Pt₂Te₂ Te O Pt bottom O Pt top (added)

Fig.1: Schematic illustration of the phase transformation of PtTe₂ to Pt₂Te₂ by adding Pt-atoms. PtTe₂ has a 1T-TMD structure and the Pt₂Te₂ phase has two layers of Pt in between the tellurium layers. For partial transformation of a PtTe₂ layer, two kinds of phase-boundaries between the PtTe₂ and Pt₂Te₂ phases can be formed along the zig-zag direction of the 1T-structure. These two phase-boundaries are labeled 'A' and 'B'.

PtTe₂ was grown by molecular beam epitaxy. This approach has been shown by several groups to result in high quality PtTe₂ films down to monolayer thickness if grown at low (250°C) temperatures.^{24,25} Growth at temperatures above 300°C favors formation of bilayer thick islands.²⁶ Figure 2 shows the as grown PtTe₂ film, that consists of both mono- and bi-layer islands. The Pt-4f XPS spectrum shows a single doublet, which is consistent with a single PtTe₂ phase sample.¹⁹ The sample was then exposed to Pt-vapor at 200°C. The results of the STM and XPS measurements with increasing Pt-exposure are also shown in Fig. 2. In the STM images, the Pt₂Te₂ phase can be distinguished from the PtTe₂ phase by a difference of their layer height as indicated in the line profiles. The Pt₂Te₂ phase is imaged as being 0.2-0.3 nm thicker than the PtTe₂ phase. In addition to the contrast change in STM images the nucleation and growth of Pt₂Te₂ by reaction of PtTe₂ with Pt, can also be followed in XPS. The two phases have distinct Pt-4f peaks with a chemical shift of ~1.2 eV between the two phases.¹⁹ Fig. 2 also shows the variation of the Pt-4f components with increasing exposure. It is apparent that the component assigned to PtTe₂ decreases while the Pt₂Te₂ component increases indicating the transformation of PtTe₂ to Pt₂Te₂. However, the transformation is incomplete and some PtTe₂ remains on the surface. From STM images it is apparent that the monolayer has completely transformed so that the remaining PtTe₂ observed in XPS is likely due to second layer PtTe₂ in the bilayer regions. The STM images also reveal the nucleation of Pt clusters on the surface with increasing Pt-deposition. Initially these clusters are primarily present in between Pt-telluride islands on the bare substrate. Once the Pt-telluride has transformed to Pt₂Te₂, Pt-clusters also nucleate

on the Pt₂Te₂ terraces. This suggests that vapor deposited Pt preferably incorporates into the PtTe₂ phase and only nucleates Pt-clusters once all the PtTe₂ is converted into Pt₂Te₂.



Fig. 2: Sample evolution in STM and XPS with increasing Pt-deposition. STM images and indicated line profiles show the conversion of $PtTe_2$ into Pt_2Te_2 . The deposition of Pt on the sample is indicated on the top. Pt was deposited at a rate of ~ 0.01 nm/min. The as grown $PtTe_2$ sample exhibits both mono- and bilayer $PtTe_2$ islands that convert into Pt_2Te_2 as Pt is deposited. Pt_2Te_2 phase exhibits slightly higher contrast as evident in the STM images and highlighted by the line profile. In the line profiles the different phases are indicated by light blue for $PtTe_2$ and light pink for Pt_2Te_2 regions. For higher Pt-exposures, Pt clusters are observed to form on the surface and the bare substrate region between the Pt-telluride islands. The conversion of $PtTe_2$ to Pt_2Te_2 is also observed in the Pt-4f core levels measured by XPS and show at the bottom of the image for different Pt-exposures. The Pt-4f is fitted with two doublets with the color of the components corresponding to Pt in a PtTe₂ environment (blue) and Pt_2Te_2 environment (pink).

First-principles calculations support the observed transformation processes. Excess Pt adsorbed on the surface of PtTe₂ strongly interact with the system giving rise to the destabilization of the Pt-Te bonds in the PtTe₂ phase (shown in Fig. S1). Such a distortion of PtTe₂ by excess Pt shows that Pt will strongly react with the substrate and form 'distorted' structures shown in Fig. S1b. For a certain concentration of Pt added to PtTe₂, the Pt₂Te₂ phase can form, which is the lowest energy structure for this composition. Specifically, using appropriate Pt chemical potentials our calculations show that formation of Pt₂Te₂ is

favored over the nucleation of metallic Pt-clusters, as illustrated in Fig. S2. Moreover, Pt-diffusion barriers on PtTe₂ have been calculated to be around 260 meV (Fig. S1c). A low diffusion barrier facilitates the growth of the Pt₂Te₂ phase from a few nucleation sites in PtTe₂ islands. Further inspections of the STM images indicates that the Pt₂Te₂ preferentially nucleates at one edge orientations of the hexagonal PtTe₂ islands as shown in Fig. 3. For extended islands nucleation and growth of Pt₂Te₂ phases are also observed. The nucleation sites for the latter are not known, i.e., if it is homogenous or heterogenous nucleation. Nucleation density may be sensitively dependent on the kinetic parameters of deposition rate and sample temperature and these parameters have not been systematically varied in this study. The Pt₂Te₂ phase has generally a triangular shape within the PtTe₂ phase. Preferred edge nucleation and Pt₂Te₂ and PtTe₂. DFT calculations of the phase boundaries suggest that the phase boundary labeled 'B' in Figure 1 is energetically preferred and thus is expected to be observed in the experiments.



Fig. 3: STM of Pt₂Te₂/PtTe₂ phase coexistence. (a) STM image of a partially conversion of PtTe₂ into the Pt₂Te₂ phase. It is apparent that the initial PtTe₂ islands have hexagonal shapes. For monolayer 1T phase the two zigzag edges are identical if the substrate does not play an important role and thus a hexagonal shape is expected. Formation of Pt₂Te₂, however preferentially occurs at only one of the two edges. Similarly, Pt₂Te₂ embedded within the PtTe₂ phase island, form preferentially triangularly shaped Pt₂Te₂ phases. As shown in Fig. 1, two different phase boundaries exist and the preference of one phase boundary over another explains the orientation of the phase boundaries in the STM image. (b) Atomically resolved STM image of interface. (c) Simulated STM image of the interface.

Making efficient electronic contacts to 2D semiconductors remains a challenge^{27,28,29,30} with covalentlybonded in-plane side-contacts generally considered to allow for better charge injections as compared to metal top contacts, because top contacts exhibit an inevitable tunnel barrier across the van der Waals gap that increases the contact resistance.³¹ High quality covalently-bonded side contacts of metals to 2D materials are, however, much more challenging to engineer and thus have mostly been studied by theoretical calculations.³¹ Detailed atomic-scale characterization and assessments of electronic properties of such side contacts are absent. The phase transformation of $PtTe_2$ into Pt_2Te_2 described above suggests that Pt-deposition onto PtTe₂ would not result in just a top-contact to PtTe₂ but instead a transformation of PtTe₂ into Pt₂Te₂ occurs even at low deposition temperatures. This implies the formation of a side contact of semiconducting monolayer $PtTe_2$ with a metallic Pt_2Te_2 phase. The structural similarity between the two phases ensures an atomically sharp contact and absence of dislocation-type defects at the interface, as the STM images of the phase-contact indicate. The controlled formation of this phase contact enables us to study the electronic structure of such a 2D metal/semiconductor side contact. In contrast to traditional buried metal/semiconductor contacts the purely 2D lateral contact is exposed to vacuum and thus can be investigated by scanning tunneling spectroscopy. To obtain the interface band alignment, both the position of the band edges relative to the Fermi-level and the work function (WF) alignment across the interface need to be determined. The former is measured directly by I-V (dI/dV) spectroscopy, while the WF differences across the interface can be assessed by field emission resonance (FER) measurements. FER spectroscopy has been widely used to determine WF modulations in 2D monolayer materials on metal substrates, e.g., for hex-BN,^{32,33} graphene,³⁴ oxide layers,^{35,36} NaCl,³⁷ or CuN³⁸. Briefly, in FER the bias voltage between an STM-tip and the sample surface is chosen such that the Fermi-level of the STM-tip lies above the vacuum level of the sample. The drop of the potential in the tunnel junction gives rise to a triangular potential well between the tip and the sample surface with discrete image potential states that are probed in the FER spectroscopy. Local changes in the substrate WF shifts these image potential states corresponding to the WF changes and thus a shift of the resonance states is a direct measure of the WF changes along the surface. While absolute WFs are difficult to extract, because of unknown STM-tip conditions, the relative difference in the WF and the spatial extent of the WF modulation can be measured accurately and this is the critical information required for the interface band alignment.

Fig. 4 shows the dI/dV spectra and FER across a $Pt_2Te_2/PtTe_2$ interface. In dI/dV spectra the Fermi-level is the reference energy at 0 eV and thus the Fermi-level of the metal is always aligned with the Fermi-level in the semiconductor. What is important is the level of the semiconductor band edges at the metal contact

and its comparison with the band edges before contact has been made, which presumably is the same as far away from the contact. For an as grown pristine PtTe₂ film the dl/dV spectrum is shown in Fig. 4(d) as the red spectrum. The calculated local density of states near the interface is presented in Fig. S3, and it is in a qualitative agreement with the experimental results. This film is p-type with the Fermi-level ~0.1 eV above the valence band edge. This band edge position is unaltered after contacting Pt₂Te₂, indicating that there is no significant shift of the Fermi-level and thus no band bending, or Schottky contact. The field emission resonance, shown in Fig. 4 (b), on the other hand, indicate a shift of the WF across the interface by 0.2 eV. This shift is very sharp and occurs over less than a nanometer at the interface, indicating that the difference in the WF is accommodated by an interface dipole on the atomic scale. These measurements are summarized in an interface band diagram shown in Fig. 4(e). The calculated band edge positions relative to the vacuum level are shown in Fig. S4 for the two materials. Consistent with the experiment, a shift of the WF by about 0.2 eV is needed to obtain the observed band alignment.



Fig. 4: Determination of electronic structure of $Pt_2Te_2-PtTe_2$ metal/semiconductor interface by low temperature (20 K) STM/STS measurements and DFT. (a) atomically resolved STM image of interface. (b) Field emission resonance spectroscopy, showing a 0.2 eV lower WF of the semiconducting $PtTe_2$ compared to the metallic Pt_2Te_2 side of the heterojunction. (c) dI/dV spectra showing the metallic properties of Pt_2Te_2

and (d) a 0.5 eV band gap of PtTe₂. For PtTe₂ spectra of a pure PtTe₂ sample (without the presence of interfaces with Pt₂Te₂) is shown by the red spectrum, which indicates an intrinsic p-type doping level. The blue and yellow spectra are taken close and far away (~10 nm) from the interface with Pt₂Te₂. All spectra show a very similar Fermi-level position and thus indicating a flat band alignment at the interface. (e) Schematic interface electronic structure derived from the STS data. The 0.2 eV WF offset between the metal and semiconductor implies an interface dipole as indicated. (f) Charge densities derived from DFT calculations (red region = increase blue region =decrease) indicating the atomic scale of the charge redistribution. at the interface. Bader charge integration shown in (g) indicate that more electrons are 'lost' (moved to the atoms exactly at the interface) from the PtTe₂ region than for the Pt₂Te₂ interface exhibit oscillation at the interface in the metallic Pt₂Te₂ phase. These oscillations may be associated with Friedel oscillations due to screening of the charges at the interface.

In the simplest model of a metal-semiconductor band alignment, the Schottky-Mott model, both the WF and Fermi-level are aligned at the interface and the required band offset in the semiconductor to achieve this band alignment causes an extended charge depletion layer in the semiconductor. In contrast to metalsemiconductor contacts in 3D materials, the reduced charge screening in 2D materials is expected to give rise to a reduced charge equilibration process at 2D- heterojunctions and a carrier delocalization.³⁹ Nevertheless, the WF offset measured here at the interface implies an atomic scale interface dipole. Interface dipoles can be caused for various reasons.⁴⁰ Spillover of metal Bloch states into the semiconducting gap (like surface dipoles on metal surfaces) would cause a dipole opposite to the one measured and thus can be excluded as the main reason for the dipole. Therefore, an atomic-scale charge transfer at the interface is most likely associated with this dipole. To get further insight into the charge redistribution at the interface, we carried out DFT calculations of the difference in the electronic charge upon contact formation. It should be pointed out that contrary to the interface made form completely different materials, the Pt₂Te₂/PtTe₂ system can be split into separate Pt₂Te₂ and PtTe₂ semi-infinite sheets in several different ways, as shown in Fig. S5, which gives rise to different binding energies between the subsystems and different charge transfer. For each division of the system, the charge transfer is confined to a few atom spaces at the interface, as illustrated in Fig. 4 (f) for the division giving the highest energy release upon interface formation. To avoid the ambiguity related to the choice of the subsystems, we adopted another strategy to assess the charge redistribution. The Bader charges in the areas (four primitive cells for each material) close to the interface were calculated and compared to those in infinite $Pt_2Te_2/PtTe_2$ systems. The results indicated that the decrease in the electronic charge is larger in $PtTe_2$ system, giving rise to a difference of 0.9×10^{-2} e/atom, with positive charge accumulated in Pt₂Te₂, in agreement with the experimental data.

In general, excess charges at metal/semiconductor interfaces must be screened and giving rise to Friedel oscillations. In this 2D system such Friedel oscillations may be imaged directly by STM. At low bias voltages oscillations are observed in STM images in the metallic Pt₂Te₂ along the interface as shown in Fig. 4 (h) and these oscillations are consistent with the screening of the excess charges at the interface.

In conclusion, a compositional phase transformation in $PtTe_2$ is enabled by on-surface reaction with vapor deposited Pt. This implies that vapor deposition of Pt on monolayer PtTe₂ causes its transformation into metallic Pt₂Te₂ and a side-metal contact to semiconducting PtTe₂. We have shown that this configuration results in a barrier-less flat band configuration where the WF differences between metal and semiconducting part are compensated by an interface dipole at the atomic scale. Thus, the heterojunction between Pt₂Te₂ and PtTe₂ can be understood by a modified Schottky Mott rule, with the only change being to replace the WF of the free metal edge with the new effective WF. The effective interface WF is the metal WF minus the interface dipole. Similar effective WFs due to interface dipoles were invoked in metalgraphene contacts.⁴¹ Interestingly, this interface dipole is just large enough to allow for a flat band alignment and thus a barrier less metal-semiconductor junction. However, the Fermi-level of the metal lies within the semiconducting gap and thus does not pin the band edges. Thus, in applications that allow shifting of the Fermi-level in the semiconductor by gating may induce charge injection barriers. Such effects should be studied in future experiments by transport measurements and/or nano-beam photoemission studies. Enabling efficient charge injection into a material with large spin-orbit coupling like PtTe₂ is a prerequisite for making efficient spin and charge devices. The unique compositional transformation from one ordered 2D phase to another may play an important role in the development of future nano-devices. The process demonstrated here of transforming one 2D compound into another by on surface reaction with vapor deposited elements, should not be limited to the Pt-telluride system only. Other 2D materials may be transformed by similar surface reactions and thus enable a new pathway for 2D material synthesis and formation of in plane 2D heterojunctions.

Acknowledgement: Financial support through NSF grant no. 2140038 is acknowledged. In addition, we acknowledge funding from the German Research Foundation (DFG), project KR 4866/6-1, and through the collaborative research center "Chemistry of Synthetic 2D Materials" SFB-1415-417590517. We further thank the Gauss Centre for Super- computing e.V. (www.gauss-centre.eu) for providing computing time

on the GCS Supercomputer HAWK at Höchstleistungsrechenzentrum Stuttgart (www.hlrs.de) and also TU Dresden (Taurus cluster) for generous grants of CPU time.

Supporting Information:

• Computational details and methods

References:

¹ Lasek, K.; Li, J.; Kolekar, S.; Coelho, P.M.; Zhang, M.; Wang, Z.; Batzill, M. **Synthesis and Characterization of 2D Transition Metal Dichalcogenides: Recent Progress from a Vacuum Surface Science Perspective.** Surf. Sci. Rep. **2021**, *76*, 100523.

² S. Cho, S. Kim, J.H. Kim, J. Zhao, J. Seok, D.H. Keum, J. Baik, D.-H. Choe, K.J. Chang, K. Suenaga, S.W. Kim, Y.H. Lee, H. Yang. **Phase Patterning for Ohmic Homojunction Contact in MoTe₂.** Science, **2015**, *349*, 625-628.

³ Y.-C. Lin, D.O. Dumcenco, Y.-S. Huang, K. Suenaga. Atomic mechanism of the semiconducting-to-metallic phase transition in single-layered MoS₂. Nat. Nanotechnol. **2014**, *9*, 391–396.

⁴ M. Acerce, D. Voiry, M. Chhowalla. **Metallic 1T Phase MoS₂ Nanosheets as Supercapacitor Electrode Materials.** Nat. Nanotechnol. **2015**, *10*, 313.

⁵ F. Wypych, R. Schöllhorn. **1T-MoS₂, a New Metallic Modification of Molybdenum Disulfide.** J. Chem. Soc., Chem. Commun. **1992**, 1386-1388.

⁶ C.A. Papageorgopoulos, W. Jaegermann. Li intercalation across and along the van der Waals surfaces of MoS₂(0001). Surf. Sci. 1995, 338, 83-93.

⁷ Kim, J. S.; Kim, J.; Zhao, J.; Kim, S.; Lee, J. H.; Jin, Y.; Choi, H.; Moon, B. H.; Bae, J. J.; Lee, Y. H.; Lim, S. C. **Electrical Transport Properties of Polymorphic MoS**₂. ACS Nano **2016**, 10, 7500–7506.

⁸ Kappera, R.; Voiry, D.; Yalcin, S. E.; Branch, B.; Gupta, G.; Mohite, A. D.; Chhowalla, M. **Phase-Engineered Low-Resistance Contacts for Ultrathin MoS₂ Transistors.** Nat. Mater. **2014**, *13*, 1128- 1134.

⁹ Duerloo, K.-A.N.; Li, Y.; Reed, E.J. **Structural Phase Transitions in Two-Dimensional Mo- and W-Dichalcogenide Monolayers.** Nat. Commun. **2014**, *5*, 4214.

¹⁰ Wang, Z.; Sun, Y.-Y.; Abdelwahab, I.; Cao, L.; Yu, W.; Ju, H.; Zhu, J.; Fu, W.; Chu, L.; Xu, H.; Loh, K.P. **Surface-Limited Superconducting Phase Transition on 1T-TaS₂.** ACS Nano **2018**, *12*, 12619–12628.

¹¹ Arnold, F.; Stan, R.-M.; Mahatha, S.K.; Lund, H.E.; Curcio, D.; Dendzik, M., Bana, H.; Travaglia, E.; Bignardi, L.; Lacovig, P.; Lizzit, D.; Li, Z.; Bianchi, M.; Miwa. J.A.; Bremholm, M.; Lizzit, S.; Hofmann, P.; Sanders, C.E. **Novel Single-Layer Vanadium Sulphide Phases.** 2D Mater. **2018**, *5*, 045009.

¹² Sutter, E.; Huang, Y.; Komsa, H.-P.; Ghorbani-Asl, M.; Krasheninnikov, A.V.; Sutter, P. Electron-Beam Induced Transformations of Layered Tin Dichalcogenides. *Nano Lett.* **2016**, *16*, 4410–4416.

¹³ Zhu, H.; Wang, Q.; Zhang, C.; Addou, R.; Cho, K.; Wallace, R.M.; Kim, M.J. New Mo₆Te₆ Sub-Nanometer-Diameter Nanowire Phase from 2H-MoTe₂. Adv. Mater. 2017, 29, 1606264.

¹⁴ Oyedele, A.D.; Yang, S.; Feng, T.; Haglund, A.V.; Gu, Y.; Puretzky, A.A.; Briggs, D.; Rouleau, C.M.; Chisholm, M.F.; Unocic, R.R.; Mandrus, D.; Meyer III, H.M.; Pantelides, S.T.; Geohegan, D.B.; Xiao, K. **Defect-Mediated Phase Transformation in Anisotropic Two-Dimensional PdSe2 Crystals for Seamless Electrical Contacts.** *J. Amer. Chem. Soc.* **2019**, *141*, 8928–8936.

¹⁵ Li, J.; Kolekar, S.; Ghorbani-Asl, M.; Lehnert, T.; Biskupek, J.; Kaiser, U.; Krasheninnikov, A.V.; Batzill, M. Layer-Dependent Band Gaps of Platinum Dichalcogenides. ACS Nano 2021, *15*, 13249-13259.

¹⁶ Gimpl, M. L.; Nelson, C. E.; Fuschillo, N. **Some Properties of Platinum Monotelluride (PtTe).** *Amer. Mineral.* **1963**, *48*, 689–691.

¹⁷ Thomassen, L. **Ueber Kristallstrukturen einiger Binaerer Verbindungen der Platinmetalle.** Z. Physik. Chem. B **1929**, 2, 349-379.

¹⁸ Cenzual, K.; Gelato, L.M.; Penzo, M.; Parthé, E. **Overlooked Triagonal Symmetry in Structures Reported with Monoclinic Centered Bravais Lattices, Trigonal Description of Li**₈**Pb**₃, **PtTe**, **Pt**₃**Te**₄, **Pt**₂**Te**₃, **LiFe**₆**Ge**₄, **LiFe**₆**Ge**₅, **CaGa**₆**Te**₁₀, and La_{3.266}**Mn**_{1.1}**S**₆. *Zeitschr. Kristallogra.* **1990**, *193*, 217-242. ¹⁹ Li, J.; Kolekar, S.; Xin, Y.; Coelho, P.M.; Lasek, K.; Nugera, F.A.; Gutiérrez, H.R.; Batzill, M. **Thermal Phase Control** of Two-Dimensional Pt-Chalcogenide (Se and Te) Ultrathin Epitaxial Films and Nanocrystals. Chem. Mater. 2021, *33*, 8018-8027.

²⁰ Coelho, P.M.; Komsa, H.P.; Diaz, H.C.; Ma, Y.; Krasheninnikov, A.V.; Batzill, M. **Post-Synthesis Modifications of Two-Dimensional MoSe₂ or MoTe₂ by Incorporation of Excess Metal Atoms into the Crystal Structure.** ACS Nano **2018**, *12*, 3975-3984.

²¹ Batzill, M. **Mirror Twin Grain Boundaries in Molybdenum Dichalcogenides.** J. Phys.: Condens. Matter **2018**, *30*, 493001.

²² Coelho, P.M.; Komsa, H.P.; Lasek, K.; Kalappattil, V.; Karthikeyan, J.; Phan, M.H.; Krasheninnikov, A.V.; Batzill, M. Room-Temperature Ferromagnetism in MoTe₂ by Post-Growth Incorporation of Vanadium Impurities. Adv. Electr. Mater. **2019**, 5, 1900044.

²³ Karthikeyan, J.; Komsa, H.P.; Batzill, M.; Krasheninnikov. A.V. Which Transition Metal Atoms can be Embedded into Two-Dimensional Molybdenum Dichalcogenides and Add Magnetism? Nano Letters **2019**, 19, 4581-4587.

²⁴ Yan, M.; Huang, H.; Zhang, K.; Wang, E.; Yao, W.; Deng, K.; Wan, G.; Zhang, H.; Arita, M.; Yang, H.; Sun, Z.; Yao, H.; Wu, Y.; Fan, S.; Duan, W.; Zhou. S. Lorentz-Violating Type-II Dirac Fermions in Transition Metal Dichalcogenide PtTe₂. Nat. Commun. 2017, *8*, 257.

²⁵ Lin, M.-K.; Villaos, R.A.B.; Hlevyack, J.A.; Chen, P.; Liu, R.-Y.; Hsu, C.-H.; Avila, J.; Mo, S.-K.; Chuang, F.-C.; Chiang, T.C. Dimensionality-Mediated Semimetal-Semiconductor Transition in Ultrathin PtTe₂ Films. Phys. Rev. Lett. 2020, *124*, 036402.

²⁶ Zhang, L.; Yang, T.; Arramel; Feng, Y.P.; Wee, A.T.S.; Wang, Z. **MBE-Grown Ultrathin PtTe₂ Film and the Layer-Dependent Electronic Structure.** *Nanoscale* **2022**, *14*, 7650-7658.

²⁷ Kwon, G.; Choi, Y.-H.; Lee, H.; Kim, H.-S.; Jeong, J.; Jeong, K.; Baik, M.; Kwon, H.; Ahn, J.; Lee, E.; Cho, M.-H. Interaction- and Defect-Free van der Waals Contacts Between Metals and Two-Dimensional Semiconductors. Nat. Electr. **2022**, *5*, 241–247.

²⁸ Wang, Y.; Kim, J.C.; Wu, R.Y.; Martinez, J.; Song, X.; Yang, J.; Zhao, F.; Mkhoyan, A.; Jeong, H.Y.; Chhowalla, M. Van der Waals Contacts Between Three-Dimensional Metals and Two-Dimensional Semiconductors. Nature 2019, 568, 70–74.

²⁹ Liu, Y.; Guo, J.; Zhu, E.; Liao, L.; Lee, S.-J.; Ding, M.; Shakir, I.; Gambin, V.; Huang, Y.; Duan, X. **Approaching the Schottky–Mott Limit in van der Waals Metal–Semiconductor Junctions**. Nature **2018**, 557, 696–700.

³⁰ Song, S.; Sim, Y.; Kim, S.-Y.; Kim, J.H.; Oh, I.; Na, W.; Lee, D.H.; Wang, J.; Yan, S.; Liu, Y.; Kwak, J.; Chen, J.-H.; Cheong, H.; Yoo, J.-W.; Lee, Z.; Kwon, S.-Y. **Wafer-Scale Production of Patterned Transition Metal Ditelluride Layers for Two-Dimensional Metal–Semiconductor Contacts at the Schottky–Mott Limit.** Nat. Electr. **2020**, 3, 207–215.

³¹ Allain, A.; Kang, J.; Banerjee, K.; Kis, A. Electrical Contacts to Two-Dimensional Semiconductors. Nat. Mater. **2015**, *14*,1195–1205.

³² Joshi, S.; Ecija, D.; Koitz,R.; Iannuzzi, M.; Seitsonen, A.P.; Hutter, J.; Sachdev, H.; Vijayaraghavan, S.; Bischoff, F.; Seufert, K.; Barth, J.V.; Auwärter, W. **Boron Nitride on Cu(111): An Electronically Corrugated Monolayer.** Nano Lett. **2012**, *12*, 5821–5828.

³³ Zhang, Q.; Yu, J.; Ebert, P.; Zhang, C.; Pan, C.-R.; Chou, M.-Y.; Shih, C.-K.; Zeng, C.; Yuan, S. **Tuning Band Gap and Work Function Modulations in Monolayer hBN/Cu(111) Heterostructures with Moiré Patterns.** ACS Nano **2018**, *12*, 9355-9362.

³⁴ Wang, B.; Caffio, M.; Bromley, C.; Früchtl, H.; Schaub, R. **Coupling Epitaxy, Chemical Bonding, and Work Function at the Local Scale in Transition Metal-Supported Graphene.** ACS Nano **2010**, *4*, 5773–5782.

³⁵ Rienks, E. D. L.; Nilius, N.; Rust, H.-P.; Freund, H.-J. **Surface Potential of a Polar Oxide Film: FeO on Pt(111)**. Phys. Rev. B **2005**, *71*, 241404.

³⁶ Lichtenstein, L.; Heyde, M.; Ulrich, S.; Nilius, N.; Freund, H.-J. **Probing the Properties of Metal–Oxide Interfaces:** Silica Films on Mo and Ru Supports. J. Phys.: Condens. Matter **2012**, *24*, 354010.

³⁷ Ploigt, H.C.; Brun, C.; Pivetta, M.; Patthey, F.; Schneider, W.-D. Local Work Function Changes Determined by Field Emission Resonances: NaCl/Ag(100). Phys. Rev. B 2007, *76*, 195404.

³⁸ Ruggiero, C. D.; Choi, T.; Gupta, J.A. **Tunneling Spectroscopy of Ultrathin Insulating Films: CuN on Cu(100).** Appl. Phys. Lett. **2007**, *91*, 253106.

³⁹ Yu, H.; Kutana, A.; Yakobson, B.I. **Carrier Delocalization in Two-Dimensional Coplanar p–n Junctions of Graphene and Metal Dichalcogenides.** Nano Lett. **2016**, 16, 5032-5036.

⁴⁰ Linga, S.; Watkinsa, M.B.; Shluger, A.L. **Effects of Atomic Scale Roughness at Metal/Insulator Interfaces on Metal Work Function.** Phys. Chem. Chem. Phys. **2013**, *15*, 19615-19624.

⁴¹ Giovannetti, G.; Khomyakov, P.A.; Brocks, G.; Karpan, V.M.; van den Brink, J.; Kelly, P.J. **Doping Graphene with Metal Contacts.** Phys. Rev. Lett. **2008**, *101*, 026803.