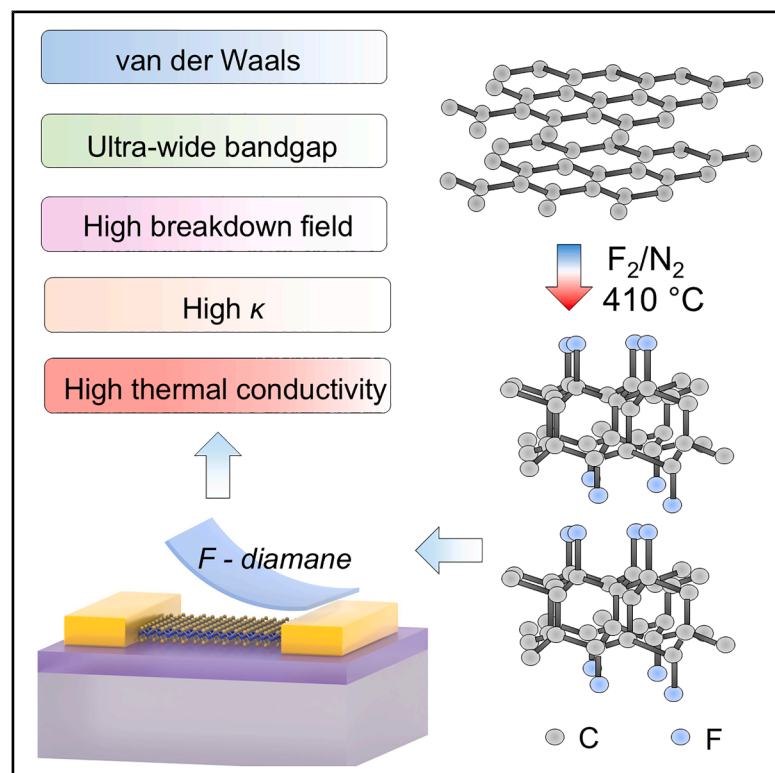


Robust van der Waals diamond-like dielectric for two-dimensional electronics

Graphical abstract



Highlights

- Fluorinated diamane is a high- κ (~ 12) van der Waals dielectric for 2D electronics
- Fluorinated diamane exhibits high breakdown field, exceeding 10 MV/cm
- Fluorinated diamane delivers high thermal conductivity (265 W/m · K) to dissipate heat
- Fluorinated diamane integrates well with a variety of 2D materials

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In brief

Next-generation two-dimensional electronics require robust insulating layers that effectively control current without compromising device performance. Here, we report fluorinated diamane, a diamond-like van der Waals material, as a high-performance gate dielectric. By combining a high dielectric constant, exceptional breakdown strength, and efficient heat dissipation, fluorinated diamane integrates seamlessly with diverse two-dimensional semiconductors and memory devices. This material platform addresses long-standing dielectric bottlenecks in ultra-scaled two-dimensional electronics.



Benchmark

First qualification/assessment of material properties and/or performance

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Article

Robust van der Waals diamond-like dielectric for two-dimensional electronics

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PROGRESS AND POTENTIAL Dielectric engineering remains a primary bottleneck in two-dimensional (2D) nanoelectronics. Conventional oxide dielectrics suffer from surface dangling bonds and interfacial disorder when integrated with 2D materials, whereas conventional 2D insulators like hexagonal boron nitride are constrained by low dielectric constants and moderate breakdown fields. This work reports the preparation of exfoliated fluorinated diamane (F-diamane), an sp^3 -bonded diamond-like material, as a robust van der Waals (vdW) gate dielectric. Synthesized via mechanical exfoliation of near-perfect $(C_2F)_n$ fluorinated graphite, F-diamane combines an ultrawide band gap (~ 5.5 eV), a high dielectric constant (~ 12), an ultrahigh electrical breakdown field exceeding 10 MV/cm (up to 20 MV/cm), and excellent thermal conductivity (~ 265 W/(m · K)). When integrated into 2D field-effect transistors, F-diamane yields high-performance p -type (WSe_2) and n -type (MoS_2) channels. Furthermore, F-diamane is demonstrated in non-volatile α - In_2Se_3 ferroelectric semi-conducting transistor with long retention times and high-mobility graphene transistors. This work establishes F-diamane as a multifunctional vdW dielectric platform that simultaneously solves key dielectric challenges in 2D electronics, such as low permittivity, dielectric breakdown, and localized thermal management. Scalable synthesis of high-quality F-diamane films could accelerate the industrial transition toward high-density, energy-efficient 2D semiconductor integration.

SUMMARY

Diamond possesses an ultrawide band gap, exceptional thermal conductivity, and outstanding dielectric strength, yet the absence of an atomically thin analogue has long hindered its integration into van der Waals (vdW) nanoelectronics. Here, we demonstrate a two-dimensional (2D) diamond-like dielectric, fluorinated

diamane (F-diamane), that preserves sp^3 -bonded carbon framework while providing atomically flat, chemically robust vdW surfaces. Through a combination of structural, electrical, thermal characterizations, we show that exfoliated F-diamane exhibits a wide band gap of ~ 5.5 eV, a large dielectric constant of ~ 12 , an ultrahigh breakdown field exceeding 10 MV/cm (up to 20 MV/cm), and excellent thermal conductivity of ~ 265 W/(m · K). Leveraging these properties, we demonstrate F-diamane as a high-quality dielectric for high-performance devices incorporating multiple classes of 2D materials, including graphene semimetal, transition-metal-chalcogenide semiconductors, and ferroelectric semiconductors. Our results establish F-diamane as a robust, multifunctional vdW dielectric and highlight its potential to overcome dielectric bottleneck in next-generation 2D electronic and memory technologies.

INTRODUCTION

Diamond, a well-known carbon allotrope with sp^3 bonding character, is renowned for its exceptional physical and chemical properties, including the ultrawide band gap (~ 5.5 eV), extremely high breakdown field (>20 MV/cm), and outstanding thermal conductivity ($\sim 2,000$ W/(m · K)), making it one of the most robust insulating materials known.^{1–5} However, integration of diamond with other semiconducting materials into nanoscale electronic devices is fundamentally hindered by its three-dimensional nature and associated fabrication constraints.⁶ Transforming diamond into layered forms has, therefore, attracted considerable interest,^{7,8} motivated not only by fundamental curiosity but also by the prospect of realizing atomically thin diamond layers that retain or even surpass some of bulk diamond's advantageous properties, while offering improved scalability and compatibility with semiconductor technologies.^{7–11} Various strategies have been explored to fabricate ultrathin diamond membranes.^{6,12} For example, Li et al. experimentally prepared the high-quality two-dimensional (2D) diamond with distinguished properties by laser-heating few-layer graphene in a diamond anvil cell.⁷ However, releasing these membranes from their growth substrates typically requires sophisticated procedures such as slicing,^{13,14} etching,¹⁵ or edge-exposed exfoliation,¹⁶ and the resulting membranes are non-van der Waals (vdW) in nature. In contrast, diamane, featuring a bilayer of sp^3 -bonded carbon, represents the ultimate 2D form of diamond-like materials.^{11,17} Early theoretical predictions indicate that AB-stacked bilayer graphene can undergo a phase transition to diamane upon surface functionalization with hydrogen or fluorine, driven by the formation of interlayer C–C bonds that transform the planar sp^2 hexagonal network into a buckled sp^3 tetrahedral lattice analogous to the (111) plane of cubic diamond.^{18–21} These processes typically require significant external pressure to initiate the phase transformation.²² Following this, a decisive experimental breakthrough was achieved by Bakharev et al. who demonstrated that controlled fluorination of chemical-vapor-deposition-grown bilayer graphene on metallic substrate induces a chemically driven conversion into a stable fluorinated diamane (F-diamane) monolayer in localized region of the bilayer graphene.²³ However, locally transforming bilayer graphene into F-diamane and the requirement of metallic substrates hinder its application in practical devices. In a more recent work, F-diamane-like structures have also been realized through fluorination of graphite to poly (dicarbon monofluoride) (C_2F)_n, followed by liquid-phase exfoliation, yielding scaled-up ultrathin

nanosheets.²⁴ Since this structure offers a chemical stable, atomically thin sp^3 carbon platform with properties fundamentally distinct from graphene, it motivates exploration beyond structural novelty toward functional roles in nano-electronic applications (Figure 1A).

In parallel, the rapid development of 2D semiconductors has opened a new vista for ultra-scaled and energy-efficient electronic systems.^{25,26} As device dimensions approach the atomic limitation, electrostatic control increasingly depends on the properties of the dielectric layer, which governs carrier confinement, interface quality, power consumption, and long-term reliability. However, dielectric integration has emerged as a central bottleneck for 2D electronics.²⁷ Conventional amorphous oxides, such as SiO₂, Al₂O₃, and HfO₂, suffer from intrinsic incompatibilities with atomically thin semiconducting materials such as transition metal dichalcogenides (TMDs).²⁸ Dangling bonds, interfacial disorder, and remote phonon scattering introduce severe mobility degradation, threshold voltage instability, and undesirable hysteresis in 2D semiconductor channels.^{29,30} vdW dielectrics, most notably hexagonal boron nitride (hBN), overcome these issues by providing atomically flat, dangling-bond-free interfaces and have become indispensable for high-performance 2D electronics.^{31–34} Nevertheless, hBN exhibits a relatively low dielectric constant (~ 4) and limited breakdown fields (~ 2 – 12 MV/cm),^{35,36} constraining further vertical scaling and operating voltage reduction. Despite considerable effort,^{37–41} the realization of a vdW dielectric that simultaneously offers a high dielectric constant, large breakdown strength, efficient thermal transport, and interfacial compatibility remains a challenge.

Here, we demonstrate that 2D diamond-like exfoliated F-diamane functions as a high-performance vdW dielectric that can be utilized in 2D electronics. F-diamane was synthesized via fluorination of natural graphite, and optothermal Raman thermometry revealed the high thermal conductivity of approximately 265 W/(m · K) near room temperature. The dielectric constant of F-diamane was determined to be ~ 12 extracted from both parallel-plate capacitors and transistors. A high breakdown field over 10 MV/cm (up to 20 MV/cm) was achieved. Using F-diamane flakes as the gate dielectric, we fabricated a high-performance *p*-type monolayer WSe₂ field-effect transistor (FET) and an *n*-type monolayer MoS₂ FET. The WSe₂ FET exhibits a room temperature hole mobility of 44.6 cm²/Vs and a minimum subthreshold swing (SS) of 112 mV/dec. The MoS₂ FET shows a room temperature electron mobility of 55.4 cm²/Vs. We also incorporate F-diamane with α -In₂Se₃ and graphene to construct a non-volatile device with a memory retention time

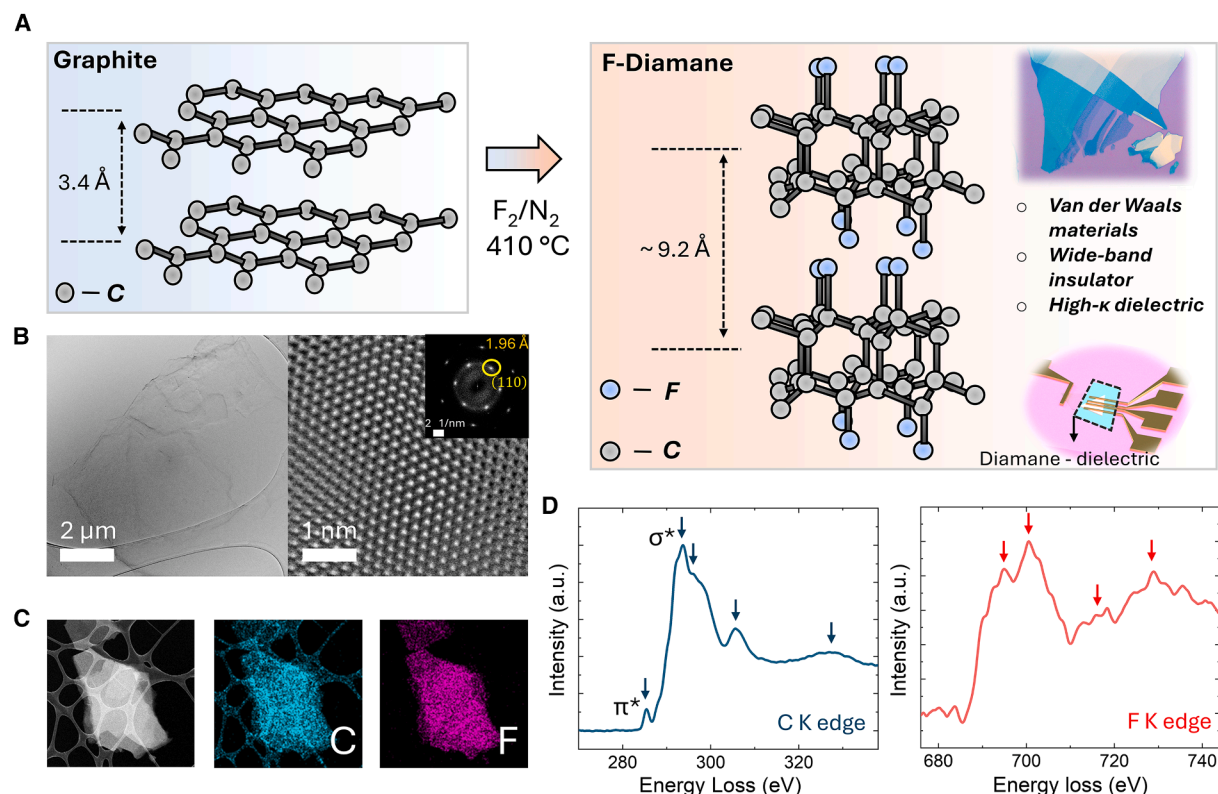


Figure 1. F-Diamane as a dielectric layer for 2D electronics

(A) Schematic of the transition from layered graphite to layered F-diamane and its application as a dielectric in 2D electronic devices.
(B) BF-TEM image showing the large-area thin flakes of F-diamane (~52.8 nm), and the HR-TEM image exhibiting the characteristic atomic arrangement. Insert is Fourier diffractogram of the HRTEM image.
(C) Elemental mapping of carbon (C) and fluorine (F) from the exfoliated F-diamane flake (left, STEM image).
(D) EELS spectra of C-K and F-K edges from the flake shown in (B).

exceeding 10^3 s and a high-mobility graphene transistor. These results demonstrate the compatibility of F-diamane to be integrated with a wide range of 2D materials for construction of next-generation 2D electronic devices.

RESULTS AND DISCUSSION

Synthesis and structural characterization of F-diamane

The near-perfect (C₂F)_n flakes were synthesized using the gas-phase fluorination method (more details are provided in [methods](#)) and used as the precursor to prepare F-diamane layers. The transmission electron microscopy (TEM) image of an exfoliated F-diamane flake, with an estimated thickness of ~52.8 nm⁴² (Figure 1B, left), confirms large-area (>8 μm) thin sheet, with the high-resolution TEM (HR-TEM) imaging and the corresponding Fourier diffractogram exhibiting the characteristic hexagonal atomic arrangement (Figure 1B, right). The interplane spacing of ~1.96 Å is assigned to be {110} planes and is consistent with previously reported TEM results of F-diamane.²³ Scanning transmission electron microscopy (STEM) elemental mapping confirms the uniform presence of carbon and fluorine across the exfoliated F-diamane flake (Figure 1C). We further characterized the spatial uniformity of the exfoliated F-diamane

flake by Raman spectroscopy mapping. Raman spectroscopy mapping was performed on the region defined by the red dashed box in Figure S1A. The resulting integrated intensity map, shown in Figure S1B, displays remarkable intensity uniformity across the outlined area, indicating the long-range homogeneity of the F-diamane structure. The TEM image taken at the folded edges of an exfoliated flake (Figures S2A and S2B) shows an interlayer distance of ~0.9 nm (Figure S2C), which is consistent with the structural model of F-diamane.⁴³ The interlayer distance ranged from 0.8 to 0.9 nm for a thicker flake (Figures S2D–S2F). This interlayer distance is further corroborated by powder X-ray diffraction (XRD), shown in Figure 2A, where the (001) diffraction peak yields an interlayer spacing of 9.2 Å.⁴³ The remaining peaks arise from the (*hk*0) planes: (100), (110), and (200) at 2.18, 1.28, and 1.09 Å, respectively. The interplanar distances were calculated based on the Bragg's law (details are provided in [Note S1](#)). Notably, no diffraction lines associated with the 00 ℓ peaks of the perfluorographane (CF)_n phase are observed (the (001) reflection would be expected at approximately 6 Å). Collectively, the powder XRD results provide clear evidence for a (C₂F)_n-rich phase. In addition, the single-crystal XRD has been conducted. Figure S3A shows the 2D image collected in coupled 2θ/ω method (horizontal axis), in which the vertical axis demonstrates

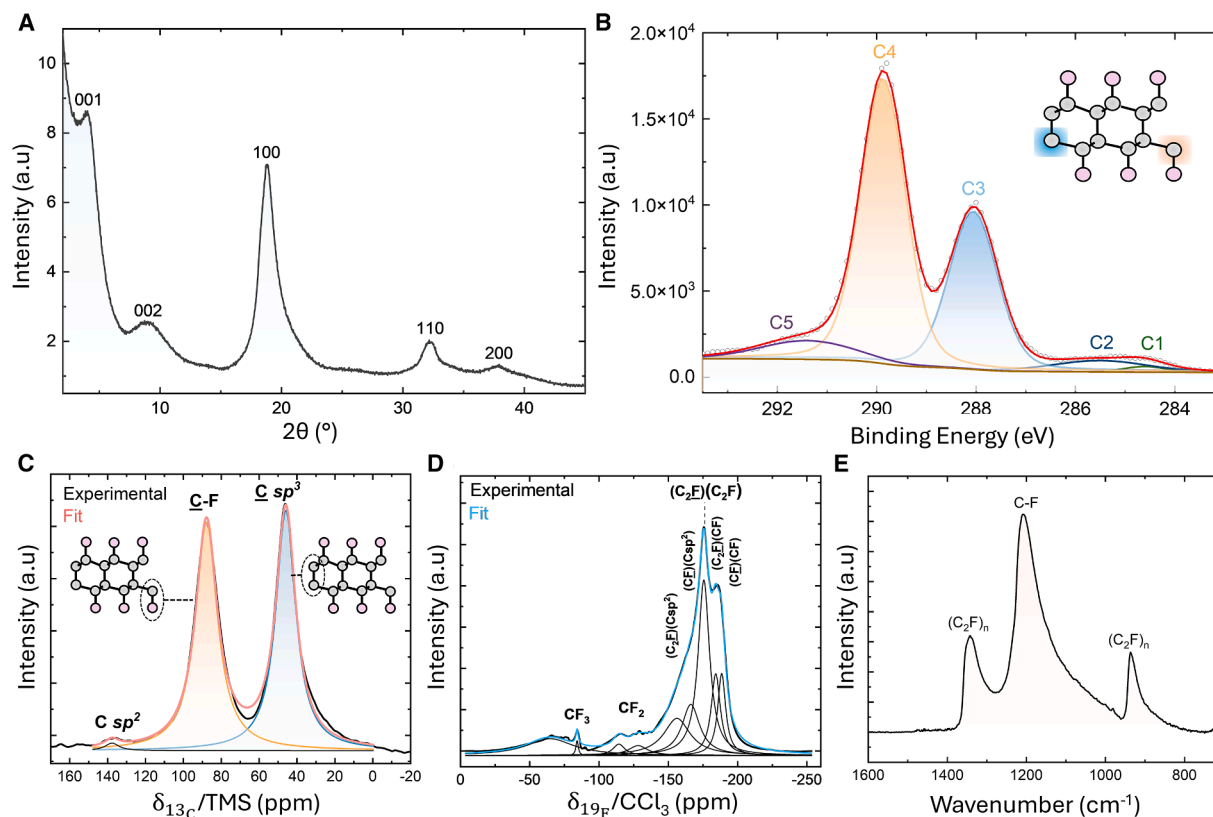


Figure 2. Chemical and structural characterizations of F-diamane

(A) X-ray diffraction pattern of F-diamane, showing characteristic diffraction peaks indexed to the (001), (002), (100), (110), and (200) planes. (B) High-resolution C 1s X-ray photoelectron spectroscopy spectrum of F-diamane. (C) ^{13}C MAS-NMR, where distinct resonances corresponding to C-F and sp^3 carbon are observed, with negligible sp^2 carbon contribution. (D) ^{19}F magic-angle spinning nuclear magnetic resonance spectrum of F-diamane. (E) Fourier transform infrared spectrum of F-diamane.

the information along χ (tilting in the plane perpendicular to the X-ray beam path) direction. Figure S3B integrates the signal along χ direction of Figure S3A and shows the 1D XRD pattern of the specimen. The broadening of (001) patterns aligns well with the powder XRD pattern (Mo K α radiation) in Figure 2A. The (001) diffraction peak (Figure S3A) also yields an interlayer spacing of 9.2 Å, which is consistent with the previous result. The single-crystal specimen contains very low graphite residual, as indexed in Figure S3B. Figure S3C shows a larger reciprocal space area by stitching multiple χ step (from -5° to 65°) 2D patterns, in which one can see that the (001) patterns become more diffuse. The thickness of the analyzed crystal sample is less than 50 μm , which is much smaller than the X-ray penetration depth. Figure S3C shows no other high-crystalline features except for the F-diamane and minor graphite residual. Therefore, we confirm that the sample has no other impurity phases. Figure 1D shows the electronic structure of F-diamane measured using electron energy loss spectroscopy (EELS) from the flake shown in Figure 1C. It is apparent that the exfoliated F-diamane maintains the characteristic diamond C-K edge features, including a pronounced intensity dip around 302.4 eV, which resembles that of diamond associated with the absolute energy gap in the

conduction-band structure of diamond.⁴⁴ The pre-edge peak at 285.4 eV indicates the minor π^* contributions from the graphitic region.⁴⁵ The additional near-edge features of multiple shoulders from 289–292 eV is consistent with the C-F bonds in graphitic fluorides.⁴⁶ The upper shift of dominating σ^* peak at 293.7 eV, followed by the broader shoulder at 297 eV, is consistent with the previous experimental and theoretical report of F-diamane.²³ The EELS F-K edge from the same region shows multiple features, starting from ~ 687 eV with a dominating peak at 700 eV, and extended fine structure to 730 eV. These multiple peaks are due to the F bonded with sp^2 and sp^3 and some fractions of disordered carbon.

X-ray photoelectron spectroscopy (XPS) was employed to investigate the surface composition of the synthesized F-diamane. The survey spectrum (Figure S4) indicates that the material is predominantly composed of carbon (61.15%) and fluorine (38.85%). The slightly higher fluorine-to-carbon ratio as compared with that from ideal F-diamane (C_2F)_n may arise from structural defects that enable edge carbon atoms to be fluorinated, forming CF_2/CF_3 species, or from the presence of a minor fraction of perfluorographane (CF)_n domains.^{47,48} In addition, the direct fluorination proceeds from the surface inward, and minor

fluorine-rich (C_2F_n) domains form on the particle surfaces. These outer regions are preferentially detected by surface-sensitive XPS, leading to an overestimation of the fluorine content compared with the true bulk composition. The high-resolution C 1s spectrum of F-diamane is shown in Figure 2B. The C1 peak at 284.5 eV is a characteristic of sp^2 carbon atoms in the graphitic domain. The C2 (285.4 eV) and C3 (288.1 eV) peaks are attributed to non-fluorinated carbon atoms adjacent to fluorinated carbons within the sp^2 and sp^3 bonding configurations, respectively. The C4 peak at 289.8 eV corresponds to carbon atoms directly bonded to fluorine (C–F), while the C5 peak at 291.5 eV is assigned to CF_2/CF_3 groups located at the peripheries of amorphous regions and/or along platelet edges.⁴⁹ Collectively, this spectral decomposition supports the formation of fluorinated diamond-like carbon structure, consistent with the chemical characteristics expected of F-diamane. In particular, the almost negligible spectral weight of sp^2 carbon (C1 and C2) demonstrates the effective transformation of graphite into F-diamane.

In agreement with previous reports for (C_2F_n) graphite fluoride,^{50–53} the ^{13}C magic-angle spinning (MAS) solid-state NMR spectrum of the fluorinated graphite exhibits two dominant resonances with nearly equal integrated intensities at 42 and 84 ppm. These are assigned to carbon atoms bonded to fluorine through covalent C–F bonds (labeled C–F in Figure 2C) and to non-fluorinated sp^3 -hybridized carbons (C–C–F), respectively. They are present only in the F-diamane (C_2F_n) phase. The very low intensity of the line at 135 ppm suggests the presence of a very small amount of residual sp^2 -hybridized carbon atoms, consistent with the XPS and XRD results. Furthermore, deconvolution of the spectra shows that approximately 94% of the C–F bonds belong to the (C_2F_n) structural type, with only 6% associated with (CF)_n domains, according to the integrated surface of the lines S_{C-F} and S_{Csp3} , with the formula $100 \times (C_2F_n)/(CF)_n = 100 \times S_{Csp3}/S_{C-F}$. Moreover, the F/C ratio can be extracted from the fits with the formula $F/C = S_{C-F}/(S_{C-F} + S_{Csp3})$. A value of 0.51 ± 0.01 is obtained, i.e., $CF_{0.51}$ or $C_2F_{1.02}$, indicating the overestimation of XPS results. Figure 2D presents the assignments of ^{19}F MAS-NMR spectra for the (C_2F_n).⁵⁴ Two types of fluorine nuclei are present, C–F and C_2F , in (CF)_n and (C_2F)_n structures, respectively. So, five lines are related to different neighborings of (CF)_n-type, (C_2F)_n-type, and non-fluorinated sp^2 C: with increasing chemical shift, (CF)–(CF) exhibits a resonance at –190 ppm, (C_2F)(CF) at –183 ppm, ($(C_2F)(C_2F)$) at –178 ppm, (CF)(Csp^2) at –168 ppm, and (C_2F)(Csp^2) at –160 ppm. Two other weak resonance lines are related to CF_2 at –120 ppm and CF_3 groups at –80 ppm. Nevertheless, the spectrum is dominated by the line at –178 ppm ($(C_2F)(C_2F)$) related to (C_2F_n) phase.

The infrared (IR) spectrum of the (C_2F_n) phase is shown in Figure 2E. The vibration of covalent C–F bonds was observed at around $1,210\text{ cm}^{-1}$. The bands at $1,340$ and 937 cm^{-1} are strongly related to the (C_2F_n) phase. A linear relationship was found between the intensity ratio, I_{1340}/I_{1210} or I_{937}/I_{1210} , and the content of (C_2F_n) phase calculated from ^{13}C MAS NMR data for a set of graphite fluorides.⁵⁴ IR spectroscopy confirms the synthesis of a quasi-perfect (C_2F_n) phase. Such a (C_2F_n) precursor with a large crystal size (200–400 μm of lateral size) is ideal for the preparation of F-diamane by mechanical exfoliation, as

shown in Figure 1A, and F-diamane can be easily exfoliated using the standard blue tape technique (Figure S5). The flakes exfoliated from the near-perfect (C_2F_n) were utilized in the following characterizations.

Lastly, the surface roughness of exfoliated F-diamane flakes was investigated using atomic force microscopy (AFM). Two representative flakes with thicknesses of approximately 40 and 30 nm exhibit root-mean-square roughness values of 0.95 and 0.97 nm, respectively (Figure S6). The consistently low roughness across flakes of different thicknesses indicates a uniform and smooth surface morphology, which is highly desirable for dielectric layers in 2D electronic devices, as it minimizes scattering induced by interfacial disorders or charged traps at the semiconductor-dielectric interface.

Dielectric properties of F-diamane flakes

Density functional theory (DFT) calculation was utilized to study the electronic band structures of F-diamane. As illustrated in the inset of Figures 3A and S7, F-diamane exhibits an intrinsic direct band-gap nature across various numbers of layers. The calculated band gap value for a monolayer F-diamane is approximately 4 eV, confirming its ultrawide-band-gap character. Then, we employed UV-visible spectroscopy to experimentally determine the optical band gap of F-diamane. Because our DFT results indicate that F-diamane is an intrinsic direct band-gap material, the experimental absorption data were evaluated using the direct transition Tauc relation $(\alpha h\nu)^2 = B(h\nu - E_g)$, where α is the absorption coefficient and E_g is the band-gap energy. By plotting $(\alpha h\nu)^2$ as a function of photon energy ($h\nu$), a sharp, well-defined fundamental absorption edge is resolved in the high-energy region (Figure 3A). Extrapolating the linear regime of this direct Tauc plot to the abscissa yields an experimental optical band gap of ~ 5.5 eV, which is comparable to that of bulk diamond. This value is larger than DFT-calculated results⁵⁵ due to well-known tendency of the generalized gradient approximation in the Perdew-Burke-Ernzerhof (GGA-PBE) functional to underestimate band gaps. A broad sub-band-gap absorption tail is observed in the lower-energy region between 2 and 5.5 eV. This tailing is attributed to localized mid-gap defect states arising from fluorination. By strictly isolating the steep fundamental edge from this defect-mediated tail background, we confirm the high wide-band-gap nature of the pristine F-diamane layers. We further assessed its insulating properties by measuring the in-plane conductivity of an individual flake. As shown in Figure S8, the extremely low source-drain current (10^{-11} A) indicates the high resistance of the flake, and its channel resistance remains unchanged under SiO_2 back-gate modulation (Figure S8B), confirming its robust insulating behavior. Then, we characterized the leakage current and breakdown field by measuring the F-diamane sandwiched in a capacitor configuration (few-layer graphene/F-diamane/Au). The leakage current density remained lower than the low-power limit until approaching the breakdown field over 10 MV/cm (~ 11.5 MV/cm for 25-nm-thick F-diamane, ~ 17.0 MV/cm for 30-nm-thick F-diamane, and ~ 20.1 MV/cm for 38-nm-thick F-diamane) (Figure 3B), in which the breakdown field of the 38-nm-thick F-diamane is close to the one of the bulk diamonds.⁴ The leakage characteristics of F-diamane follow the Frenkel-Poole (FP) mechanism (more details are provided in Note S2). The plots in Figure 3C exhibit clear

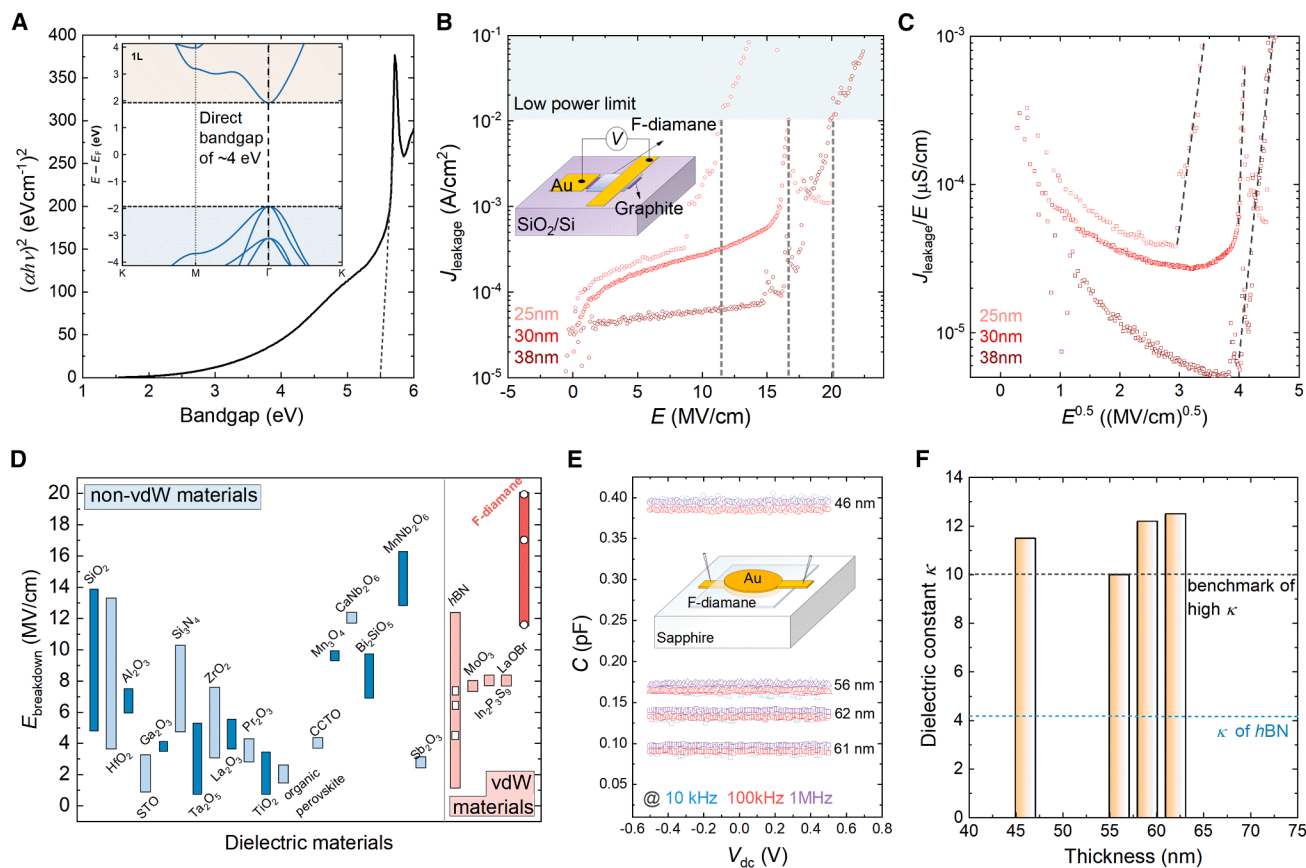


Figure 3. Dielectric properties of F-diamane flakes

(A) Tauc plot for F-diamane flakes obtained from UV-visible spectroscopy to determine the band gap (optical). Insert represents the electronic band structure of monolayer F-diamane by DFT calculation.
 (B) Electrical breakdown field measurements of F-diamane based on metal-insulator-metal capacitors.
 (C) Frenkel-Pool plot of the high-field F-diamane conductivity as a function of electric field.
 (D) Comparison of the breakdown field of F-diamane with representative dielectric materials. Individual data points for the hBN controls (white squares) and F-diamane capacitors (white circles) are superimposed on their respective bars.
 (E) Capacitance measurements based on F-diamane capacitors at different thicknesses.
 (F) Extracted dielectric constant (κ) for F-diamane flakes at different thicknesses. The black dash line represents the high- κ benchmark, and the blue dash line represents the measured dielectric constant of the control device, Au/hBN/Au capacitor.

linearity, giving slopes of 6.7, 8.9, and 6.9 for 25-, 30-, and 38-nm-thick F-diamane flakes, which are consistent with the theoretical FP coefficient for a dielectric constant of 11 (this is the experimental value from Figure 3F, which will be discussed later). This agreement indicates that field-assisted thermal excitation from trap states dominates the leakage process in exfoliated F-diamane flakes. Compared with control few-layer graphene/hBN/Au capacitors (Figure S9), F-diamane flakes exhibit a markedly higher breakdown field. Figure 3D summarizes the breakdown field for a wide range of other dielectric materials, categorized into non-vdW and vdW classes (Table S1). It should be noted that F-diamane stands out with the highest breakdown field, highlighting its exceptional dielectric robustness and great potential in power electronics. We further characterized the dielectric constant by measuring several parallel-plate capacitors (Au/F-diamane/Au metal-insulator-metal [MIM] architecture) (Figure 3E). Across different thicknesses, the measured capacitance re-

mained stable, showing no significant fluctuations. The dielectric constant of F-diamane was determined using $\kappa = \frac{C \times d}{A \times \epsilon_0}$, where C is the measured capacitance of the dielectric layer, A is the effective area of the capacitor, d is the thickness of the F-diamane flakes, and ϵ_0 is the vacuum permittivity. The thickness of the flakes was determined using AFM (Figure S10). The extracted values were nearly constant across all thicknesses, varying only slightly between 10 and 12.5. Notably, these values are approximately 3-fold greater than those of hBN extracted from a control device (Figure S9D), confirming the high- κ characteristic of F-diamane flakes (Figure 3F). In addition, we notice that this dielectric constant (κ) is two times larger than that of diamond (approximately 5.7).⁵⁶ This enhancement primarily originates from the strong polarity of the C–F bonds, which introduces substantial ionic polarization absent in diamond. In addition, the reduced dimensionality and surface termination in F-diamane

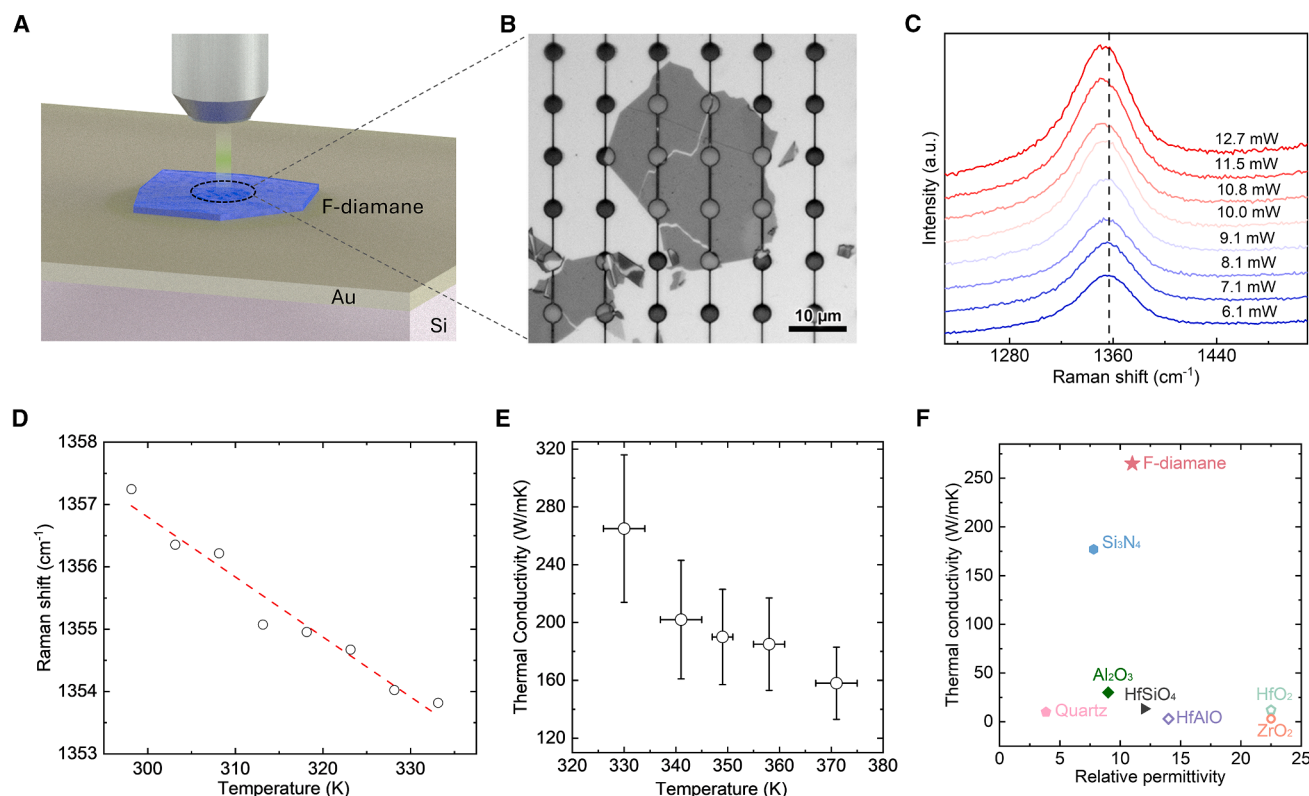


Figure 4. Thermal conductivity of exfoliated F-diamane

- (A) Schematic of the experimental setup for optothermal Raman measurement.
 (B) Optical image of suspended F-diamane flake on an Au-coated Si substrate with patterned microwells.
 (C) Shift of the Raman peak as a function of laser power for a suspended F-diamane flake.
 (D) Temperature-dependent shift of the Raman peak of F-diamane under controlled heating.
 (E) Experimental thermal conductivity of F-diamane as a function of temperature. The uncertainty calculation is provided in [Note S3](#).
 (F) Comparison of the thermal conductivity and dielectric constant of F-diamane with those of representative industrial dielectric materials.

allow larger lattice and dipolar responses under an external electric field, further increasing the dielectric constant.^{57,58}

Thermal conductivity of F-diamane flakes

In nanoscale electronics, the thermal conductivity of 2D dielectrics is also a crucial factor, as efficient heat dissipation prevents local overheating and maintains device performance and reliability, highlighting the importance of high- κ materials with good thermal transport.⁵⁹ To quantify the thermal transport properties of exfoliated F-diamane, we employed optothermal Raman thermometry.^{60,61} In our device configuration (Figures 4A and 4B), an F-diamane flake (~ 11 nm determined by AFM) was suspended over a SiO₂/Si wafer featuring prefabricated microwells (diameter: 3.8 μm), which were connected via 200-nm-wide trenches (Figure S11). These trenches mitigated pressure-induced deformation of the suspended film during laser heating, while a 100-nm-thick sputtered gold film served as heat sinks. Figure 4C illustrates the evolution of F-diamane's Raman spectra with increasing laser power from 6.1 to 12.7 mW, which induced a red-shift of the diamond-like *sp*³ peak. This Raman peak of F-diamane differs from the T_{2g} mode of pristine diamond ($\sim 1,332$ cm⁻¹ at 300 K). We attribute this discrepancy to F-dia-

mane's 2D lattice nature and reduced crystal symmetry, which alter the local bonding environment and force constants of the *sp*³ carbon network. Figure 4D displays the temperature-dependent Raman response of F-diamane, showing the systematic shift of the *sp*³-related peak and the corresponding linear fits used to extract the first-order temperature coefficient. Assuming the heat flux vector is along the radial direction away from the center of the suspended F-diamane sheets and phonon transport is diffusive, the thermal conductivity was obtained using Equation 1, with the resulting temperature-dependent values plotted in Figure 4E.

$$\gamma_{\text{thermal conductivity}} = \frac{\ln\left(\frac{R}{r_0}\right)Q}{2\pi t(T_m - T_a)}\alpha, \quad (\text{Equation 1})$$

where R is the radius of the microwell with the value of 1.9 μm; r_0 is the radius of the laser beam with the value of 0.28 μm; α is the Gaussian profile factor of the laser beam with the value of 0.97; t is the thickness of the F-diamane flake determined using AFM; T_m is the temperature measured by Raman; T_a is the ambient temperature; and Q is the absorbed laser power (Figure S12).

At near room temperature, F-diamane exhibits a thermal conductivity of ~ 265 W/(m·K). The systematic uncertainty

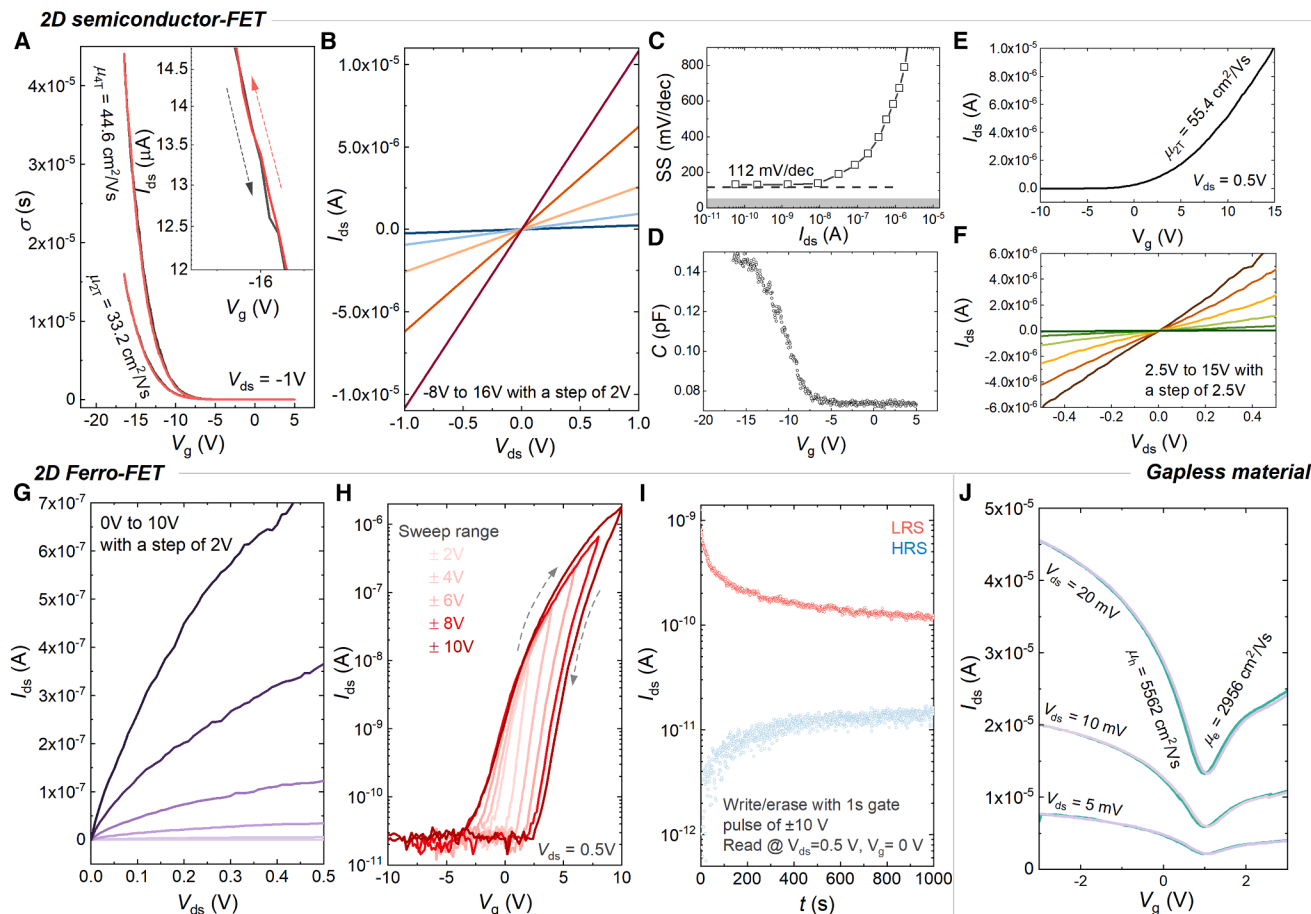


Figure 5. 2D electronic devices employing F-diamane as the gate dielectric

(A) Two-terminal and four-terminal transfer curves for monolayer WSe₂ FET, using F-diamane as the dielectric. Insert shows negligible hysteresis.
 (B) Output curves of the monolayer WSe₂ FET.
 (C) Subthreshold swing of the monolayer WSe₂ FET.
 (D) Capacitance-voltage (C-V) measurement for the monolayer WSe₂ FET.
 (E) Two-terminal transfer curves for the monolayer MoS₂ FET.
 (F) Output curves of the monolayer MoS₂ FET.
 (G and H) Output curves and transfer curve hysteresis of α -In₂Se₃ FeS-FET with various gate voltage (V_g) sweep ranges.
 (I) Retention characterization of α -In₂Se₃ FeS-FET by gate pulse control.
 (J) Transfer curves of the graphene device at different source-drain voltages (V_{ds}).

calculation is provided in [Note S3](#). The thermal conductivity substantially exceeds that of conventional dielectric materials used in industrial semiconductor processing ([Figure 4F](#)). This underscores its potential as a heat-spreading dielectric for next-generation 2D electronics.

Device application with F-diamane dielectric

To investigate the dielectric performance of exfoliated F-diamane in 2D electronics, we integrated it as a dielectric layer in devices based on three representative classes of 2D materials: semiconductors, ferroelectrics (FEs), and semimetals.

TMD-based p- or n-type FETs

A four-terminal monolayer WSe₂ FET (back-gate architecture: WSe₂/F-diamane/graphite) was first studied at room temperature (fabrication details and device architecture are described in

[methods](#)). [Figure 5A](#) shows both two-terminal and four-terminal transfer curves (conductance as a function of gate bias) with a constant V_{ds} of -1 V . The device exhibits typical *p*-type enhanced-mode behavior. The fitted two-terminal mobility is $33.2 \text{ cm}^2/\text{V}\cdot\text{s}$, and the four-terminal FE mobility is $44.6 \text{ cm}^2/\text{V}\cdot\text{s}$; these values are determined by $\mu_{2T} = \left(\frac{1}{C_{ox} \cdot V_{ds}} \right) \left(\frac{L}{W} \right) \frac{\partial I_{ds}}{\partial V_{gs}}$ for the

two-terminal measurement and $\mu_{4T} = \frac{1}{C_{ox}} \cdot \frac{D}{W} \cdot \frac{\partial (I_{ds}/V_{4T})}{\partial V_{gs}}$ for the four-terminal measurement. In these equations, C_{ox} is the calculated capacitance of F-diamane (dielectric constant is adopted according to previous capacitance measurements), L and W represent the channel length and width, respectively, and D is the distance between the two inner electrodes. These extracted four-terminal and two-terminal mobilities are close to each other, which is due to the excellent ohmic contacts with a MoO₃

interfacial layer⁶² (the symmetric and linear output curves in Figure 5B at all gate bias demonstrate the ohmic contact behavior, which is consistent with the literature). The fitted two-terminal mobilities are significantly larger than the one extracted from a controlled device (forward two-terminal mobility $\sim 1.4 \text{ cm}^2/\text{V}\cdot\text{s}$) with conventional SiO_2 /heavily doped Si back-gating. This is because the F-diamane layer provides an atomically flat and minimal-trap-state interface that minimizes carrier scattering and Coulomb impurities compared with the SiO_2 (285 nm)/Si substrate. Utilizing an identical device architecture (monolayer WSe_2 /dielectric/few-layer graphene), the monolayer WSe_2 FETs incorporating F-diamane exhibit substantial performance enhancements, in particular for the carrier mobility and SS, demonstrating the clear advantages of F-diamane over the exfoliated $h\text{BN}$ at the device level (Figure S13; Table S2). The SS (Figure 5C) is determined to be 112 mV/dec (significantly smaller than 900 mV/dec for $h\text{BN}$ device), which is attributed to the high- κ nature of F-diamane. Moreover, the high-quality interfaces are corroborated by the inset of Figure 5A, which shows that there is almost no hysteresis effect, suggesting minimal trap states in either F-diamane or the interface between WSe_2 and F-diamane, whereas the controlled device, monolayer WSe_2 /285 nm SiO_2 /Si, shows a significant large hysteresis effect (the introduction of $h\text{BN}$ dielectric mitigates hysteresis, but a residual loop persists) (Figures S13A and S14A). From the capacitance-voltage (C-V) measurement in a two-terminal WSe_2 FET fabricated on sapphire, the accumulation and depletion region can be clearly observed, as shown in Figure 5D, and the effective dielectric constant (κ) is calculated to be ~ 12 , which is consistent with the MIM measurement presented above. In addition, an n -type monolayer MoS_2 FET has been studied. The field-effect mobility was extracted to be $55.4 \text{ cm}^2/\text{V}\cdot\text{s}$ from the transfer curve (Figure 5E). Output curves show excellent linearity at all gate biases. These results collectively demonstrate that F-diamane can serve as an excellent vdW dielectric layer that preserves the intrinsic transport properties of both p -type and n -type TMDs. Consequently, these metrics position F-diamane as a promising complementary metal-oxide-semiconductor (CMOS)-compatible dielectric for future ultrathin, high-performance electronics.

$\alpha\text{-In}_2\text{Se}_3$ -based ferroelectric semiconductor FETs

Recently, 2D FE materials have emerged as promising candidates for future semiconductor technology. Unlike the traditional non-vdW FE materials, 2D FE materials maintain their ferroelectricity, even down to the monolayer limit, providing outstanding flexibility for the construction of non-volatile vdW semiconductor devices. Ferroelectric semiconductor FETs (FeS-FETs), in which 2D FE materials act both as channel materials and ferroelectric materials, have attracted enormous attention due to their non-volatile memory behavior and enhanced electrostatic control due to the intrinsic coupling between ferroelectricity and semiconducting behavior. This dual functionality results in both upward and downward band bending at the semiconductor-dielectric interface as the flip of ferroelectric polarization and sets higher requirements for compatible dielectric materials. We employed $\alpha\text{-In}_2\text{Se}_3$ ($\sim 12 \text{ nm}$), a typical 2D FE semiconductor, and F-diamane flakes to construct a back-gated FeS-FET ($\alpha\text{-In}_2\text{Se}_3$ /F-diamane/graphite). The output curve shows typical n -type behavior, as

shown in Figure 5G. The transfer curve exhibits clear clockwise hysteresis under bidirectional gate voltage scanning due to the switching of the polarization of $\alpha\text{-In}_2\text{Se}_3$, as shown in Figure 5H, which is consistent with the literature.⁶³ A high on-off ratio of over 10^4 is reached under a gate-bias sweep range of $\pm 10 \text{ V}$, with a minimal leakage current in the order of 10^{-11} A . Two distinct resistance states, labeled as low-resistance state (LRS) and high-resistance state (HRS), can be identified. We examine the non-volatile properties of FeS-FET by performing a retention characterization, as shown in Figure 5I. By applying a gate pulse of 1-s width at -10 V (10 V), the transistor was set to LRS (HRS) accordingly. Continuous measurement of the source-drain current shows a long retention time of over 1,000 s. All the above tests demonstrate that F-diamane serves as an effective dielectric material compatible with ferroelectric semiconductor materials.

High-mobility graphene transistor

Lastly, we fabricated an all-carbon-based transistor, in which graphite functions as the gate electrode, F-diamane serves as the dielectric, and exfoliated graphene acts as the active channel. The Fermi level of graphene can be tuned across the Dirac point (Figure 5J), and the extracted hole and electron mobilities reach 5,562 and $2,956 \text{ cm}^2/\text{V}\cdot\text{s}$, respectively. These exceptionally high mobilities highlight the pristine interface quality between graphene and F-diamane. Moreover, the charge-neutrality point appears near 1 V, indicating minimal charge transfer at the F-diamane-graphene interface.

Conclusion

In this work, we report exfoliated F-diamane flakes as a robust vdW dielectric suited for integration into 2D electronic devices. Exfoliated F-diamane simultaneously delivers a wide band gap of $\sim 5.5 \text{ eV}$, a high dielectric constant of ~ 12 , and an exceptional breakdown field exceeding 10 MV/cm (up to 20 MV/cm with 38-nm -thick F-diamane flakes), surpassing most reported vdW dielectrics. A thermal conductivity of $\sim 265 \text{ W}/(\text{m}\cdot\text{K})$ further distinguishes F-diamane from conventional dielectric materials by enabling efficient heat dissipation for high-performance 2D devices. When integrated into 2D devices, F-diamane preserves high carrier mobility in both monolayer WSe_2 and MoS_2 , supports stable ferroelectric switching in $\alpha\text{-In}_2\text{Se}_3$ FeS-FETs, and enables graphene transistors with ultrahigh carrier mobilities and negligible interfacial charge transfer. These results establish F-diamane as a robust, multifunctional dielectric platform for next-generation 2D electronics and all-carbon device architectures.

Beyond its exceptional performance as a passive dielectric insulator, the F-diamane flake holds significant potential as an active channel for high-power electronics.¹¹ While conventional substitutional doping remains a notorious challenge in bulk diamonds, the atomistic thickness of F-diamane lends itself perfectly to highly efficient surface transfer doping,^{11,64–66} opening new avenues for ultrathin, high-power electronic devices. To fully realize these technological promises, transitioning from exfoliated flakes to the scalable synthesis of high-quality, large-scale single-crystal films represents the next step. Recent breakthroughs in the controlled, large-scale growth of $h\text{BN}$ ⁶⁷ provide a promising roadmap for the scalable synthesis of

F-diamane, paving the way for its industrial integration into future semiconductor manufacturing.

METHODS

F-diamane formation

Following Kita et al.,⁵⁴ the fluorination of natural graphite (Sigma Aldrich, flakes of 200–400 μm lateral size and 1–150 μm thick) was performed in pure F_2 gas at a reaction temperature of 410°C in the “ $(\text{C}_2\text{F})_n$ -rich” region (380°C–450°C). An up-scalable static mode was used to perform fluorination for 48 h in a 5-L tubular reactor made of nickel alloy and passivated with NiF_2 . In this mode, fluorine was added in a closed reactor starting from primary vacuum (10^{-3} mbar) and up to 1 atm. 5 g of graphite was placed onto 2 plateaux. Before the reaction, moisture and oxygen were removed by applying a primary vacuum (10^{-3} mbar) at 200°C.

Device fabrications

Devices for thermal conductivity characterization

The trench-connected micro-wells on Si substrates were fabricated by both photolithography and electron-beam lithography with a depth of ~ 2 μm . Then Au was deposited on it by a metal sputter (Leica EM ACE600). Then the F-diamane flakes were transferred on it to form the suspended architecture by dry transfer using a poly (bisphenol A carbonate) stamp. The devices for laser absorbance measurement were fabricated by transferring the F-diamane flakes on TEM grids.

2D transistor fabrication

F-diamane and graphite flakes were mechanically exfoliated onto an SiO_2 (285 nm)/Si substrate by the blue tape (P/N: 1009R-1.0) technology at room temperature separately, as shown in Figure S15A. The chosen F-diamane flake was picked up by a poly(bisphenol A carbonate) (PC) stamp, then aligned and laminated onto the target graphite to form the back gate architecture. The sacrificed PC film was washed away by soaking in CHCl_3 for 5 min at 60°C. Then, TMD (MoS_2 , WSe_2 and $\alpha\text{-In}_2\text{Se}_3$) flakes were mechanically exfoliated onto a polydimethylsiloxane (PDMS) stamp. The monolayer nature of MoS_2 and WSe_2 was determined by the optical contrast and photoluminescence measurements (Figure S16). After annealing the back-gate heterostructures in Ar ambient (500 sccm Ar at 300°C) for 2 h to eliminate residual PC, the TMD flakes were transferred on the back-gate heterostructure, using the dry transfer technology (Figure S15B). Lastly, the electrodes were fabricated using standard lithography (e-beam lithography for four-terminal devices, and photolithography for two-terminal devices), e-beam evaporation, and a lift-off process (Figure S15C).

Capacitor fabrication

For breakdown field measurements, exfoliated F-diamane flakes (or *h*BN flakes) were aligned and transferred on the few-layer graphene by the standard PC method. Then, the electrodes were fabricated by the standard photolithography, e-beam evaporation, and lift-off processes. For dielectric constant extraction, bottom electrodes (Ti/Au) were fabricated on the sapphire substrates by the standard photolithography technique. Then, exfoliated F-diamane flakes (*h*BN flakes for control devices) were picked up and aligned to the bottom electrode by us-

ing the standard PC method. Lastly, the top electrodes were fabricated on the transferred F-diamane flakes (or *h*BN flakes) to achieve the MIM structure.

Chemical and structural characterization

X-ray diffractograms

X-ray diffractograms were recorded on a Bruker D8 Advance diffractometer using a molybdenum anticathode ($\lambda_{\text{K}\alpha 1} = 0.70930$ Å) and equipped with a 2D LynxEye detector and a 2.5° divergence Soller slit. The recorded spectra were accumulated for 1 h in a 2°–45° (2θ) angular range, repeated 8 times in order to increase the signal-to-noise ratio. Samples were placed in Lindemann glass capillaries (1.5 mm diameter) and rotated at 10 rpm in order to minimize the preferential orientation.

The single-crystal XRD has been conducted using a Rigaku SmartLab diffractometer equipped with Cu $\text{K}\alpha$ radiation ($\lambda = 1.54$ Å). The beam is conditioned using a multilayer mirror (CBO) to parallel beam, then using a poly-capillary focusing unit (CBO-f) to point beam with ~ 0.5 -mm radius. The single-crystal specimen was fixed at the goniometer center and aligned using the X-ray beam. The detector was calibrated in the 2D imaging mode to collect the diffraction signal.

Nuclear magnetic resonance spectra

^{19}F and ^{13}C solid-state NMR spectra were acquired on a Bruker AVANCE spectrometer. 2.5- and 4-mm zirconia rotors rotating at 30 and 10 kHz for ^{19}F and ^{13}C nuclei, respectively, were used. For MAS spectra, a simple sequence was performed with a single $\pi/2$ pulse length of 4.0 and 3.5 μs for ^{19}F and ^{13}C , respectively. ^{13}C NMR was performed at a frequency of 73.4 MHz, and tetramethylsilane (TMS) was used as the reference. ^{19}F NMR was carried out with a frequency of 282.2 MHz, and the spectra were externally referenced to CFCl_3 .

Fourier transform infrared spectra

Fourier transform infrared (FTIR) spectra were recorded on a Nicolet 6700 FT-IR (Thermo Scientific) spectrometer in the transmission mode at 4 cm^{-1} resolution, and 256 scans were taken for each spectrum.

XPS spectra

XPS spectra were recorded using an ESCALAB 250 Xi (Thermo Scientific) with a monochromated Al $\text{K}\alpha$ X-ray source (1,486.68 eV) and a pass energy of 100 eV for survey scans or 20 eV for region scans.

TEM imaging and spectroscopy

TEM specimens were prepared by firstly sonicating the F-diamane in chloroform for 15 min, followed by drop-casting the dispersion onto lacey carbon film-coated Cu grids (SPI Supplies) and air drying. The STEM-EDX analysis was performed using a JEOL JEM-F200 Multi-Purpose FEG-S/TEM operating at an accelerating voltage of 200 kV. The aberration-corrected TEM imaging and EELS were carried out using JEOL GrandArmF2 equipped with a direct electron camera (K3, Gatan) at EMU, UNSW. EELS was done with a dispersion of 30 meV and an energy resolution of 0.4 eV.

Thermal conductivity measurements

Raman measurements were performed using a Renishaw inVia system with a 488 nm excitation laser and a 100 $\times$ objective lens (NA = 0.90). The spectra were calibrated against the Si reference peak at 520.5 cm^{-1} . The laser power delivered through the

objective was determined using a Newport 1916-C optical power meter, and temperature control during the measurements was provided by a Linkam LTS350 heating stage. For the laser absorbance of the F-diamane measurement, the F-diamane film was transferred onto a SiN_x TEM grid, and the light absorbance was calculated from the power difference between the TEM grid holes covered with F-diamane and those uncovered.

Electrical characterization

Electrical transport measurements

Electrical transport characterizations of the TMD (monolayer WSe₂ and MoS₂) and graphene transistors were performed in a home-built vacuum chamber. Two Keithley 2400 source meters were employed to apply the source-drain voltage (V_{ds}) and gate voltage (V_{gs}) and monitor the source-drain current (I_{ds}) and gate leakage current. A voltage meter (Agilent 34461A) was used to measure the V_{xx} through the inner probes of the four-terminal device. The Keysight B1500A semiconductor device parameter analyzer integrated with a probe station was employed to process the C-V measurements. Electrical characterizations of the FeS-FET were performed using a home-built probe station integrated with a Keithley 2636B.

Optical characterization

Photoluminescence measurement

A continuous-wave laser at 532 nm was employed for excitation. A 20× objective lens (numerical aperture, 0.5) was applied to focus the laser down to the diffraction limit. The optical power on the sample was kept below 100 μW to avoid heating and degradation. The photoluminescence signal was filtered by a long-pass filter (Semrock LP03-532RE), dispersed by a grating-based monochromator (Horiba, iHR-550), and detected by a nitrogen-cooled charge-coupled device (Horiba, Symphony).

DFT calculation details

The electronic band structure of F-diamane is calculated using spin-polarized DFT calculations implemented in the Vienna *ab initio* simulation package (VASP)⁶⁸ with the projector augmented wave (PAW) method and the GGA-PBE form.⁶⁹ A plane-wave cutoff energy of 500 eV was adopted. A 15 × 15 × 1 gamma-centered k-point mesh is used for Brillouin zone sampling.

RESOURCE AVAILABILITY

Lead contact

Requests for further information and resources should be directed to and will be fulfilled by the lead contact, Qingdong Ou (qdou@must.edu.mo).

Materials availability

The materials generated in this study may be made available upon request.

Data and code availability

The data that support the findings of this study are available from the [lead contact](#) upon reasonable request.

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AUTHOR CONTRIBUTIONS

Conceptualization, K.X., S.C., D.-C.Q., and Q.O.; methodology, K.X., P.Z., Q.C., Y.I.C., and J.H.; investigation, K.X., P.Z., Q.C., W.Z., Y.Y., S.W., J.B., M.D., H.W., and S.L.Y.C.; visualization, K.X., P.Z., Q.C., M.D., S.C., and Y.Y.; supervision, K.X., S.C., D.-C.Q., and Q.O.; writing – original draft, K.X. and P.Z.; writing – review & editing and discussion, K.X., P.Z., Q.C., W.Z., Y.Y., H.W., S.L.Y.C., J.L., S.W., J.B., J.H., M.T.E., C.G., X.Y., M.S.F., M.D., Y.I.C., S.C., D.-C.Q., and Q.O.

DECLARATION OF INTERESTS

The authors declare no conflict of interest.

SUPPLEMENTAL INFORMATION

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