

Wafer-scale epitaxy growth of high-mobility p-type boron carbon nitride

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A long-standing bottleneck in realizing two-dimensional (2D) CMOS technology lies in the lack of high-performance p-type semiconductors^{1–4}. Strong electron-doping tendencies, orbital localization and pronounced hole scattering have collectively impeded the development of stable, efficient p-type 2D materials. Here we report the epitaxy growth of boron carbon nitride (BCN) as a high-performance p-type semiconductor. By engineering the dehydrogenation and surface reaction pathways of monomethyl ammonia borane (MMAB) and ammonia borane (AB), we overcome a crucial barrier: the spatial and temporal mismatch in the delivery of boron (B), carbon (C) and nitrogen (N) atoms, which disrupts lattice uniformity. The result is a wafer-scale, monolayer 2D BCN in which C atoms and dimers primarily substitute for N sites within a continuously crystallized, locally distorted boron nitride lattice, leading to a sizable bandgap of 1.90 eV. Wafer-scale arrays of p-type BCN FETs exhibit benchmark performance, with a field-effect hole mobility of $100 \text{ cm}^2 \text{ V}^{-1} \text{ s}^{-1}$, on-current $>0.9 \text{ mA } \mu\text{m}^{-1}$, on–off ratio of 10^8 and threshold voltage of -0.45 V , surpassing current state-of-the-art p-type 2D semiconductors. Our findings establish BCN as a scalable and stable p-type platform, bridging a critical gap in the materials palette for three-dimensional monolithic integration of complementary electronics.

Despite considerable progress in n-type two-dimensional (2D) semiconductors, the development of high-performance p-type 2D counterparts is challenging^{1–4}. Challenges such as strong electron doping, orbital localization and excessive hole scattering have severely limited the number, stability and transport performance of p-type 2D materials, preventing their practical and monolithic integration into energy-efficient three-dimensional (3D) complementary field-effect transistors (CFETs)^{5,6}. Although the discovery of graphene in 2004 ignited hopes for revolutionary electronics, its lack of a bandgap remains a key limitation. This led to a flourishing body of research on 2D transition-metal dichalcogenides (TMDs), which offer tunable electronic properties and enable new devices⁷. However, most TMDs exhibit n-type or ambipolar conduction because of interfacial charge impurities and structural defects, unless they are carefully engineered at the atomically thin van der Waals (vdW) gap between TMDs and metal electrodes⁶, geometrically confined within the dielectric domains^{8,9} or meticulously hyperdoped by gate-driven type-III band modulation¹⁰.

On the other end of the periodic table, p-type 2D non-TMD semiconductors, including single elements (black phosphorus¹¹), ternary compound (graphene-hybridized boron nitride¹²), chalcogenides (Te) and oxides ($\beta\text{-Bi}_2\text{O}_3$, ref. 13, and Se-alloyed TeO_2 , ref. 14), tend to exhibit carrier mobilities one to two orders of magnitude lower than their n-type counterparts, coupled with poor stability and limited scalability under processing and operational conditions. These performance disparities in n-type and p-type transport properties, uniformity, scalability, stability and lack of industrial compatibility inherently hinder the tantalizing prospect of implementing these materials as p-type channels¹⁵.

In this study, we propose a wafer-scale epitaxy growth strategy for one-atom-thick, p-type semiconducting boron carbon nitride (BCN) with high mobility and industrial compatibility. As a ternary 2D material, BCN represents a unique expansion of the hexagonal boron nitride (hBN)^{16–18} and graphene families^{19,20}, integrating properties from both extremes of the electronic spectrum. Although graphene is a semimetal with exceptional carrier mobility, hBN is a wide-bandgap

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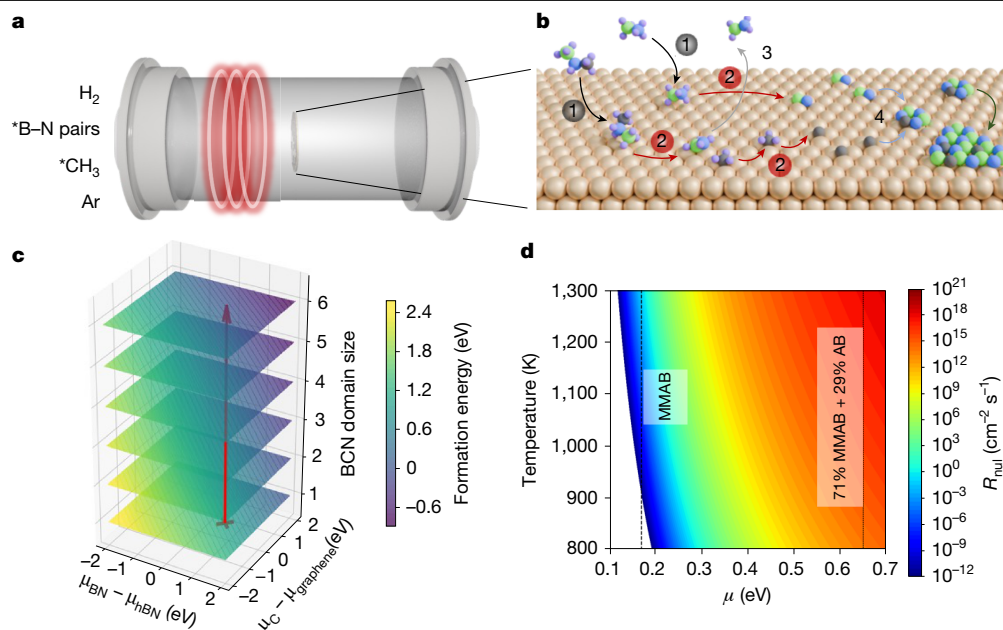


Fig. 1 | Simulation and wafer-scale synthesis of BCN. **a**, Schematic of our customized CVD setup. AB and MMAB are introduced separately into the growth setup using individual MFCs. **b**, Reaction Gibbs free energy results shed light on the possible reaction pathways on Cu (111), in which adsorption of C^* and subsequent coupling with B-N^* into BCN are thermodynamically favoured. To this end, the formation of BCN occurs through (1) MMAB and AB adsorption; (2) dehydrogenation and N-C bond cleavage; (3) H_2BNH_2^* desorption; (4) B-N^* and C^* nucleation; and (5) finally, coalescence into BCN. **c**, Formation energy of

BCN nuclei ($L = 1-6$) on Cu (111) compared with $\Delta\mu_{\text{BN}}$ (referenced to hBN) and $\Delta\mu_{\text{C}}$ (referenced to graphene). Larger domains are progressively stabilized, yielding a wider region of negative formation energy consistent with thermodynamically favourable lateral growth under C- and BN-rich conditions. The 'X' symbol represents the optimized chemical environment, and the arrow indicates the continuous improvement in stability during growth. **d**, 2D nucleation rate map further underscores the marked difference in nucleation rates (about $10^{15} \text{ cm}^{-2} \text{ s}^{-1}$) when using the binary precursors of AB and MMAB.

insulator (5.9 eV). Theoretical studies have long predicted that BCN exhibits p-type transport characteristics, high carrier mobility and direction-dependent thermal management if atomic-scale hybridization among B, N and C can be achieved^{12,21}. At the heart of the challenge is the uncontrollable, rampant hybridization of graphene-like domains (for example, local, C-rich motifs such as monomers, dimers or trimers), embedded within hBN lattices, which competes with well-defined heteroatom substitution on a similar energy scale²². This is because of the high activation energy for triple- and double-bond conversion or C-H bond cleavage, either of which must occur before C-B or C-N bond formation can take place²³. Furthermore, the spatial and temporal mismatch in the delivery of B, C and N atoms hinders the formation of BCN lattices with a heteroatom substitution. To this end, we engineer the dehydrogenation and surface reaction pathways by designing a dual-precursor system, that is, a single-molecule precursor, monomethyl ammonia borane (MMAB, C-bearing), coupled with ammonia borane (AB, BN supplier). We propose that synchronized surface delivery of BN- and C-containing fragments is essential for achieving atomically substituted BCN and for suppressing phase-separated graphene-like domains. Next, 2D potential energy surface (PES) simulations map key reaction coordinates, thereby identifying energy barriers, selectivity windows and bifurcation points that determine the kinetics and thermodynamics of each transformation step on the substrate. This enables sequential dehydrogenation of C-containing fragments, followed by subsequent coupling with BN pairs in customized low-pressure chemical vapour deposition. The result is the monolayer (unless otherwise stated) 2D BCN with atomically substituted C embedded between the continuously crystallized but locally distorted BN lattices. Our multi-scale characterizations provide a deterministic structure-property-performance relationship that links atomic-scale bonding to device-level performance in wafer-scale FET arrays. The results are a mobility (μ_{eff}) of about $100 \text{ cm}^2 \text{ V}^{-1} \text{ s}^{-1}$, an on-current (I_{on}) of about $0.9 \text{ mA } \mu\text{m}^{-1}$, a threshold voltage (V_{th}) of around -0.45 V and

an on-off current ratio ($I_{\text{on}}/I_{\text{off}}$) of approximately 10^8 . Moreover, we demonstrate uniform growth of monolayer 2D BCN on 3D-patterned SiO_2/Si substrates, thereby facilitating future integration into complex device architectures.

Engineering precursor pathways

Figure 1a and Supplementary Fig. 1a show the dissimilar reaction pathways of MMAB and AB as dual precursors on Cu (111). First-principles calculations of reaction energies of each elementary step at 1,293 K (Supplementary Fig. 1b,c) show that AB preferentially undergoes direct dehydrogenation to form surface-bound B-N^* pairs (Fig. 1b, (1)). The decrease in energy at high temperature is consistent with a previous study²⁴. Notably, although the direct dehydrogenation pathway of MMAB exhibits the lowest overall energy change, energy decomposition analysis shows that cleavage of the C-N bond requires less reaction energy²⁵ (Supplementary Fig. 2a). This suggests that MMAB is more likely to decompose by the formation of an H_2BNH_2^* and CH_3^* intermediate pair (Fig. 1b, (1) and (2)). Following the desorption of H_2BNH_2^* (Fig. 1b, (3)), the remaining CH_3^* undergoes further sequential dehydrogenation steps before active C atoms are available for BCN nucleation (Fig. 1b, (4) and (5), and Supplementary Fig. 2b). Compared with CH_4 , the decomposition of MMAB involves additional bond-breaking steps. This creates a 'kinetic delay' in the release of atomic C. From a kinetic standpoint, this delay effectively lowers the instantaneous concentration of free C atoms on the Cu surface and prolongs the time required for C to aggregate. Because surface diffusion and C aggregation are kinetically inhibited by this staged release, the formation of isolated graphene-like domains is significantly suppressed. This 'slow-release' mechanism allows the C atoms to intermix more effectively and co-nucleate with the B and N species provided by the AB precursor, leading to a chemically dispersed BCN unit, which subsequently nucleates and coalesces into BCN flakes, rather than phase-separated graphene and hBN²¹.

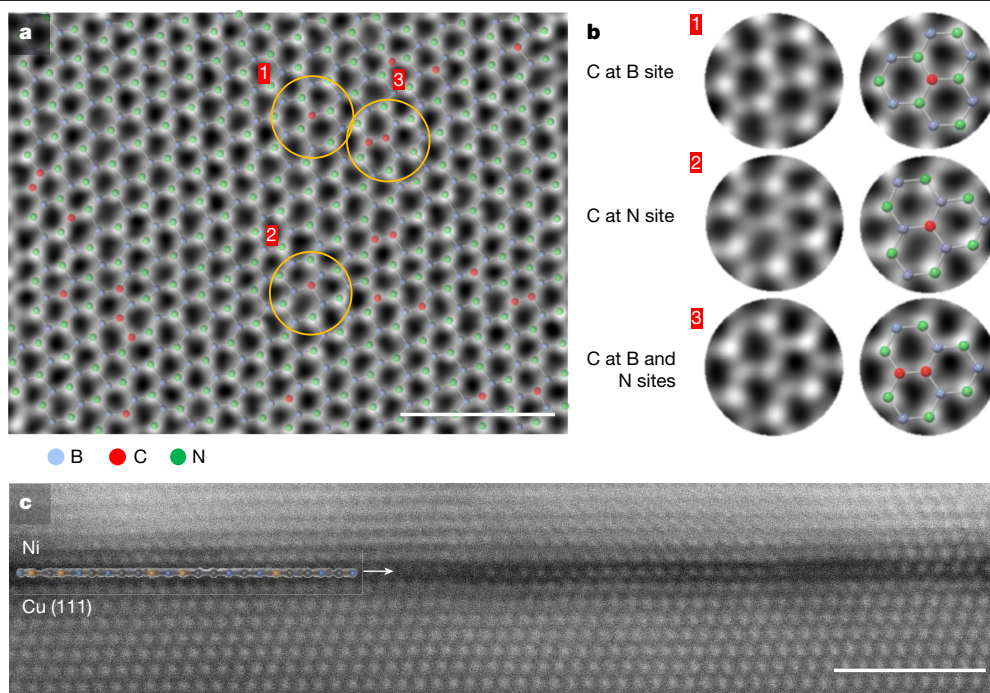


Fig. 2 | Atomically resolved HR-STEM images of monolayer BCN film. **a**, An HR-STEM image of monolayer BCN film. The exact coordination environment of each atom can be resolved. Atom-by-atom analyses show that 16 C atoms, both in the atomically substituted form and in dimers, are found at the N site, whereas 9 are found at the B site, as indicated in areas 1, 2 and 3. Note that, despite the long-range order and uniformity, the hexagonal lattices are locally

distorted, closely resembling the strained silicon (Si). **b**, Atomically substituted C and dimers are the prevailing forms of doping within the continuously crystallized but locally distorted BN lattices. **c**, Cross-sectional HR-STEM image indicates that the as-grown BCN is one atomic layer thick. Scale bar, 1 nm (**a**); 2 nm (**c**).

Figure 1c presents the calculated formation energy of BCN nuclei ($L = 1-6$) on Cu (111), mapped compared with $\Delta\mu_{\text{BN}}$ (referenced to hBN) and $\Delta\mu_{\text{C}}$ (referenced to graphene). Larger domains are progressively stabilized, yielding a wider region of negative formation energy consistent with thermodynamically favourable lateral growth under C- and BN-rich conditions. Full computational details are provided in the Supplementary Notes. The PES reveals an apparent trend towards stabilization with increasing BCN size. Moreover, when the chemical potential of the precursors exceeds that of hBN and graphene, the formation of BCN flakes becomes thermodynamically favoured. Conversely, a lower precursor chemical potential leads to a hybridized phase composed of hybridized hBN and graphene domains. The 'X' symbol in Fig. 1c denotes the chemical potential coordinates (μ_{MMAB} and μ_{AB}) corresponding to the optimized experimental precursor ratio used in our study. Furthermore, this coordinate indicates the unique thermodynamic environment in which the synergistic effect of the dual-precursor system stabilizes the formation of the monolayer BCN relative to the separate graphene and hBN phases. The upward arrow emanating from this 'X' symbol illustrates the growth path and the progression of stability of the BCN domains under these constant chemical potential conditions.

Epitaxial growth of BCN on Cu (111)

To further explain the kinetics of BCN nucleation, a 2D nucleation rate map on Cu (111) is shown in Fig. 1d (see also Supplementary Fig. 3 and Supplementary Methods). Based on classical nucleation theory, the results show an exponential increase in nucleation rate with increasing chemical potential and temperature²⁶. Although a low chemical potential equivalent to MMAB alone ($\mu = 0.18$ eV) can facilitate BCN nucleation, the corresponding rate is several orders of magnitude lower than that achieved with an optimized MMAB: AB ratio ($\mu = 0.66$ eV), which yields nucleation rates exceeding $10^{15} \text{ cm}^{-2} \text{ s}^{-1}$. This disparity

reflects the intrinsic limitation of MMAB: its pre-existing B–N–C bonds reduce the chemical potential difference between the precursors and the BCN film, thereby weakening the thermodynamic driving force for large-area monolayer BCN growth. By contrast, proper introduction of AB raises the chemical potential of the precursor pool, restoring the strong driving force that enables efficient, large-scale BCN synthesis. With this concept in mind, we optimized the CVD growth, in which dual precursors of MMAB (71%) and AB (29%) were separately annealed in sub-chambers and subsequently diluted in hydrogen and argon as carrier gases (Methods and Supplementary Fig. 4). The concentration of each reactant can be modulated by mass flow controllers (MFCs) and deposition time, thus controlling the nucleation rate while maximizing the areal coverage^{27,28}. Specifically, an atomic force microscopy (AFM) image taken after initial nucleation further suggested that edge attachment of predominantly mono-oriented BCN flakes was the primary mechanism for monolayer growth (Supplementary Fig. 5). Figure 2 summarizes our key findings from multi-scale characterizations.

Coordination environment of BCN

Although it is possible to tune the atomic ratio of B, C and N by controlling the types and fluxes of precursors, from graphene (C, 100%) to hBN (C, 0%), the formidable challenge lies in the indeterministic control of the incorporation of C-atoms^{12,23} (Supplementary Fig. 6). To investigate the locations of single B, C and N atoms, we performed atomic-resolution scanning transmission electron microscopy (STEM), in which an atomically focused electron probe is scanned over a monolayer sample, and scattered electrons are collected to form atomic-resolution STEM images. Here, we use the high-angle annular dark-field (HAADF) STEM imaging mode, in which the contrast is strongly related to atomic number (Z), a property known as Z -contrast. In HAADF-STEM imaging, the Z -contrast scales approximately with $Z^{1.7}$, and heavier atoms show brighter contrast. As shown

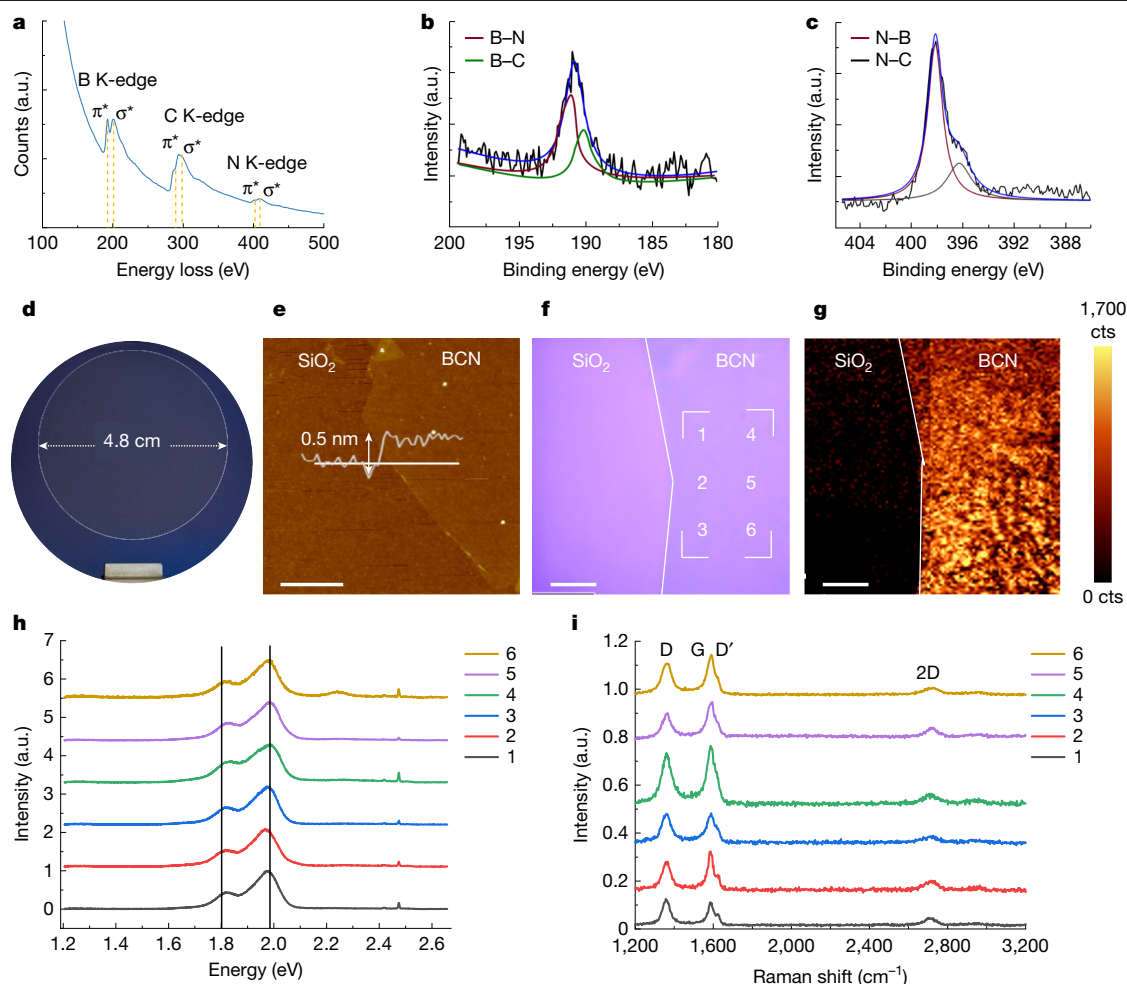


Fig. 3 | Multi-scale characterizations of BCN. **a**, K-shell excitations of B, C and N are revealed from the core EELS spectra taken from monolayer BCN films. **b,c**, XPS spectra of the B (**b**) and N (**c**) 1s core levels, in tandem with deconvoluted curves obtained by Gaussian fitting, suggest multi-bonding environments, confirming the substitution of B and N by C. **d**, A photograph shows the integrity of a monolayer BCN film transferred from the growth substrate to a Si/SiO₂

substrate. **e**, Line scan from AFM determines the step height from the edge of BCN to the supported Si/SiO₂ substrate (step height of about 0.5 nm). **f,g**, Optical microscope image (**f**) and the corresponding PL mapping (**g**) at 1.85 eV collectively attest to the spatial uniformity. **h,i**, Photoluminescence (**h**) and Raman spectra (**i**) taken at the six marked positions on the monolayer BCN film on SiO₂ (300 nm). Scale bars, 400 nm (**e**); 10 μ m (**f,g**).

in Fig. 2a, C substitutions ($Z = 6$) display lower contrast than N atoms ($Z = 7$), whereas they appear brighter than B atoms ($Z = 5$). The exact coordination environment of each atom can be resolved at the atomic level. Atom-by-atom analysis shows that 16 C atoms are found in the N sites (C at N, C_N), whereas 9 are found in the B sites (C at B, C_B), as shown in areas 1, 2 and 3 (Fig. 2b and Supplementary Fig. 7a). All C atoms are coordinated in either atomically substituted form or dimers ($C_{B\&N}$). No hybridized C domains are present, thus corroborating the efficacy of the simulation, precursor engineering and CVD synthesis. Although the overall hexagonal structure is maintained, this heteroatom doping is accompanied by substantial lattice distortions. In parallel, a cross-sectional HR-STEM image indicates that our BCN film consists of a single layer (Fig. 2c). This growth mode closely resembles that of graphene on Cu, which also yields a single layer under similar conditions. In parallel, electron energy-loss spectroscopy (EELS; Supplementary Fig. 7b) mapping shows the distribution of C atoms across the cross-section of thicker BCN films (more than five layers). It is important to note that we did not detect the presence of Cu or CuO_x, ruling out possible contamination during CVD growth. Combining HR-STEM, cross-sectional HRTEM and EELS mapping provides insights into the exact coordination environment.

To determine their chemical composition and bonding type, we conducted EELS by energy losses associated with ionization edges.

As shown in Fig. 3a, BCN spectra typically exhibit K-shell edges at 185 eV (B), 276 eV (C) and 393 eV (N)²⁹. Well-resolved 1s $\rightarrow \pi^*$ and broader 1s $\rightarrow \sigma^*$ features confirm the sp^2 hybridization of all three elements, indicating the formation of a hexagonal BCN lattice on Cu (111), consistent with the HR-STEM results in Fig. 2a (ref. 30). Complementary X-ray photoelectron spectroscopy (XPS) measurements identify the dominant bonding configurations as shown in Fig. 3b,c. The shape and position of the spectra of all three elements deviate from those of previously reported BC₂N, hybridized hBN and graphene domains. Two prominent peaks in the 1s spectrum of B can be resolved at 191.0 eV and 188.0 eV, respectively. Apart from the B 1s (191.0 eV) in hBN, the pronounced shoulder at 188 eV indicates the contribution of B and C. The N 1s peak at 398.3 eV, close to that of hBN (398.1 eV), with a higher-energy shoulder, suggests partial N–C bonding. Semi-quantitative XPS fitting of the B 1s, C 1s and N 1s core levels gives an average B:N:C ratio of about 1.14:1.06:1.01. The overestimation of C is expected because XPS is surface-sensitive and susceptible to contributions from adventitious C and the underlying substrate. Meanwhile, atom-resolved HR-STEM/EELS analysis yields B:N:C $\approx$ 51.6%:39.8%:8.6% because HR-STEM/EELS probes only the BCN lattice and thus provides a more reliable local composition. The result indicates that B remains near-stoichiometric, whereas C primarily substitutes for N sites rather than forming a separate C-rich phase. Both XPS and HR-STEM/EELS analyses consistently

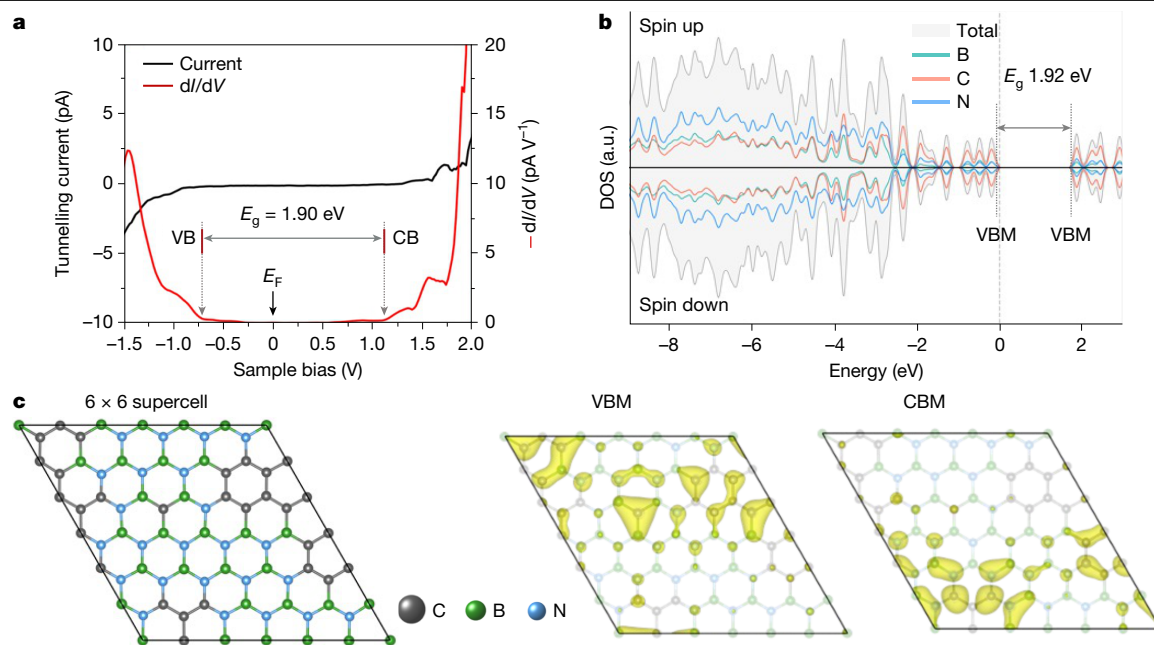


Fig. 4 | Electronic structure and band gap. **a**, STM/S measurements determine the bandgap of a monolayer BCN on Cu (111), resulting in a value of about 1.90 eV. The Fermi level is aligned to 0 eV. **b**, DOS of the monolayer BCN model supports the experimental observation, indicating a bandgap of about 1.92 eV. **c**, Partial

charge density plots of the VBM and CBM show the spatial distribution of the delocalized charge density. The isosurface value of the partial charge density is set to 0.001 e Br⁻³, a.u., arbitrary units.

show that B is slightly more abundant than N in the BCN domains, whereas C remains the minor component²⁸. This B-rich, N-deficient signature is fully consistent with C substituting N sites in the BN lattice, as supported by the observed B–C and N–C bonding configurations in the XPS spectra and by atom-resolved HR-STEM.

To evaluate the structural integrity and compositional uniformity of BCN films, Raman spectroscopy was performed on samples transferred onto a 2-inch Si/SiO₂ substrate as shown in Fig. 3d. Apart from the uniform and smooth morphology, AFM thickness measurement taken across the white solid line determines the thickness of the BCN film of about 0.5 nm (Fig. 3e). Moreover, optical microscope image and the corresponding photoluminescence mapping, featured in Fig. 3f,g, probed at 1.85 eV collectively attest to the spatial uniformity. Spectra acquired at six randomly selected locations across the film reveal four prominent features: a pronounced and broad band at 1,350 cm⁻¹, a broadened G band (around 1,588 cm⁻¹), a D' band (around 1,620 cm⁻¹) and a reduced 2D band (around 2,700 cm⁻¹; Fig. 3h,i). In traditional graphene systems, the D and D' bands are typically associated with structural defects or grain boundaries that act as scattering centres. However, in our monolayer BCN alloy, we reconcile these Raman features and the high hole mobility through the following points. The prominent D and D' bands in BCN do not necessarily represent structural damage or amorphous regions. Instead, they primarily arise from the breaking of local symmetry caused by the intentional atomic substitution of C into the hBN lattice sp^2 hybridization. Unlike random physical defects, these C substitutions are integrated into the crystalline lattice, preserving the overall sp^2 hexagonal network. To investigate the effect of these substitutions on the lattice integrity, we performed a detailed strain analysis on our monolayer BCN model (Supplementary Fig. 8). Our results indicate that the local lattice distortion caused by C incorporation is remarkably small (calculated strain <0.1%). This minimal distortion suggests that the periodic potential of the lattice remains largely intact, which is crucial for maintaining efficient carrier delocalization. Despite the chemical disorder detected by Raman spectroscopy, the electronic states in our BCN films remain highly delocalized because of the robust π -electron system of the sp^2 network³¹.

Similar to photoluminescence, the Raman spectra obtained from all six positions, as well as the large-area mapping, exhibit highly consistent peak positions, shapes and relative intensities, underscoring the spatial uniformity and low surface roughness of the BCN films (Supplementary Fig. 9). Specifically, this uniformity attests to the reproducibility of the low-pressure chemical vapour deposition growth process.

STS/M and DOS bandgap determination

Figure 4a shows the electronic properties of BCN grown on Cu (111) through constant-height scanning tunnelling spectroscopy measurements (STS/M)³². By analysing the differential conductance (dI/dV) spectrum, we identified the energy positions at which the electronic states first become available for electron tunnelling, locating the conduction band edge at 1.20 eV and the valence band edge at -0.7 eV. These measurements directly reveal a bandgap of 1.90 eV, thus confirming the semiconducting nature of the material. The resulting bandgap is slightly smaller than the one measured by photoluminescence, because STS probes the single-particle electronic bandgap by directly mapping the density of states (DOS) in real space. Nevertheless, the spectrum exhibits clear p-type behaviour, with the valence band lying closer to the Fermi level than the conduction band, which aligns well with our device-level electrical measurements (discussed in the following paragraph). Using insights from STS/M, we performed a density functional theory (DFT) calculation to investigate the electronic structure and electrical properties of BCN. The theoretical bandgap of BCN aligns well with the STS/M results, exhibiting a slightly lower bandgap of 1.92 eV (Fig. 4b). Band gap and work function variation with different BCN ratios are also presented (Supplementary Fig. 10). Moreover, the machine-learning force field combined with the Monte Carlo simulations (Supplementary Figs. 11 and 12) indicate the stable microstructure of BCN should be quasi-random, comprising isolated C and graphene-like nanodomains. The DOS shows that p-type behaviour and a narrow band gap originate from the C replacement at the isolated N site (C_N configuration) and the graphene-like nanodomain, respectively.

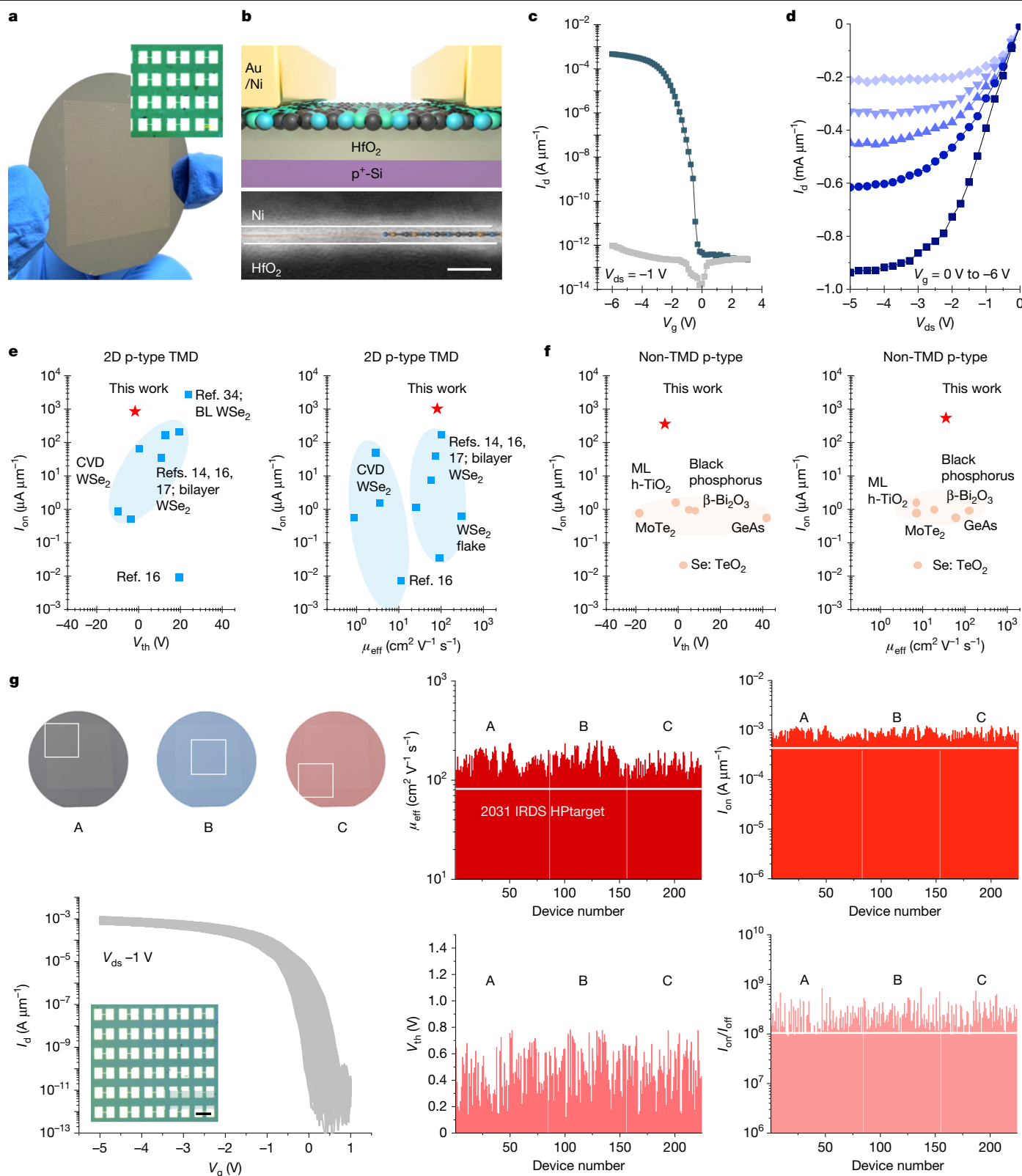


Fig. 5 | Electrical characterization of p-channel BCN FETs on a 20 nm HfO_2 . **a**, Photograph of BCN FET arrays on a 2-inch HfO_2/Si wafer. The inset shows micrographs of an individual FET and FET arrays. **b**, Schematic shows the BCN FET geometry. The corresponding cross-sectional HRTEM image shows a clean vdW gap between the BCN and Ni/HfO_2 interface. **c**, Transfer characteristics of BCN FETs at $V_{\text{ds}} = -1\text{ V}$, where L_{ch} (channel length) is $1\text{ }\mu\text{m}$. Obtained results exhibit a maximum on-current density (I_{on}) of up to $0.9\text{ mA }\mu\text{m}^{-1}$, threshold voltage (V_{th}) of -0.45 V , an average field-effect mobility (μ_{eff}) of $100\text{ cm}^2\text{ V}^{-1}\text{ s}^{-1}$,

and an on/off ratio of 10^8 . The grey line is the gate leakage current. **d**, Output characteristics of a BCN FET. **e**, Benchmark of I_{on} compared with V_{th} and I_{on} compared with μ_{eff} of reported 2D TMD p-channel FETs. **f**, Further comparisons of $I_{\text{on}}-V_{\text{th}}$ and $I_{\text{on}}-\mu_{\text{eff}}$ of the non-TMD p-channel FETs. **g**, Statistical distribution of 224 BCN FETs (across three batches) for μ_{eff} , I_{on} , V_{th} and $I_{\text{on}}/I_{\text{off}}$ ratios measured at different locations on a 2-inch wafer. The dotted lines indicate the requirements for meeting the 2031 International Roadmap for Devices and Systems challenge for high-performance targets.

We also noted that when a C substitution occurs at a C_N configuration, both the valence band maximum (VBM) and conduction band minimum (CBM) shift to higher energy levels (upward shift) relative to pristine hBN. This shift, combined with the fact that C ($2s^2 2p^2$) has one fewer valence electron than N ($2s^2 2p^3$), effectively introduces ‘holes’ into the system (Supplementary Fig. 13). The C_N sites thus act as electron acceptors. In parallel, the partial charge density of the VBM and CBM shows a delocalized charge density distribution, as shown in Fig. 4c. The DOS analysis reveals that both the VBM and CBM are predominantly derived from C atoms (Fig. 4b, orange line). This conclusion is further supported by the partial charge density distribution, which shows a clear localization of electronic states around C sites. Furthermore, the band alignment of BCN on Cu (111) is constructed by correlating the ultraviolet photoelectron spectroscopy (UPS)-derived valence band (E_v) onset, low-energy inverse photoemission spectroscopy (LEIPS)-derived conduction band (E_c) onset and the KPFM-extracted work function (E_f) of about 4.8 eV (Supplementary Fig. 14). The resulting diagram reveals a bandgap of 1.9 eV with $E_c = -3.2$ eV, $E_v = -5.1$ eV and E_f pinned close to the valence band edge, as shown in Supplementary Fig. 15. Together, these results underscore that the C_N configuration arises as the dominant structural motif governing the electronic properties of our BCN films.

Electrical performance of BCN p-type FETs

To assess the potential of BCN-based semiconductors for electronic device applications, we developed bottom-gate, top-contact p-type FETs to facilitate comprehensive evaluation. Notable electrical properties of our monolayer BCN films include several key characteristics: (1) excellent transport properties that outperform those of leading p-type semiconductors, including β - Bi_2O_3 (ref. 13), black phosphorus nanoribbons¹¹, amorphous Se-alloyed TeO_2 (ref. 14), bilayer WSe_2 (refs. 9,33), monolayer WSe_2 (refs. 34,35), monolayer h- TiO_2 (ref. 36); (2) uniformity over large areas; and (3) industrial compatibility with 3D patterned insulating substrates. We fabricated p-channel FETs featuring BCN films as channels on a 20 nm HfO_2 dielectric, with Au/Ni source and drain electrodes (0.5-nm-thick BCN FET, length $L = 1$ μm , width $W = 5$ μm), as schematically shown in Fig. 5a,b (see also Methods and Supplementary Fig. 16). Figure 5b features the cross-sectional HRTEM of the monolayer 2D BCN after depositing on HfO_2 , revealing a clean vdW gap between BCN and Ni/ HfO_2 interface. The transfer and output characteristics are shown in Fig. 5c,d. The BCN FETs demonstrated typical p-channel behaviour, with an average μ_{eff} of $100 \text{ cm}^2 \text{ V}^{-1} \text{ s}^{-1}$ (extracted from both forward and reverse scans; Supplementary Note and Supplementary Fig. 17), an $I_{\text{on}}/I_{\text{off}}$ of 10^8 , a low V_{th} of -0.45 V and an I_{on} of $>0.9 \text{ mA } \mu\text{m}^{-1}$ at room temperature. Apart from the high field-effect mobility, the I_{on} , a more direct and reliable measure for assessing transistors with 2D semiconductors, compares favourably with the bilayer WSe_2 benchmarks using VSe_2 vdW contacts³⁷. Meanwhile, the BCN FET exhibits a peak transconductance (g_m) of $325 \mu\text{S } \mu\text{m}^{-1}$ (Supplementary Fig. 18). These values are mutually consistent with the extracted field-effect mobility of about $100 \text{ cm}^2 \text{ V}^{-1} \text{ s}^{-1}$ and reflect the transporting ability and gating efficiency of BCN, even without comprehensive contact optimization. From the output curves, good current saturation (linearity) was observed at high (low) source–drain voltage, ruling out the influence of charge defects at the contact interface⁶. In parallel, a slightly negative V_{th} ensures that the device turns on reliably when the gate is pulled to ground (0 V), while the source is at a positive supply voltage (for example, 3.3 V or 5 V). Achieving high-performance p-type 2D TMD devices necessarily involves evaporating high-work-function metals onto atomically thin 2D TMDs or tailoring carrier density at the atomically thin limit while preserving an ultraclean vdW interface between the 2D TMDs and 3D metals. Nonetheless, the overall electrical properties of FETs made with BCN exceed those of the best reported properties for single-crystalline monolayer/bilayer WSe_2 -based

FETs derived from exfoliated or CVD samples with p-type electrical contacts (Fig. 5e,f; see Supplementary Tables 1 and 2 for details). Meanwhile, FETs comprising thin films derived from 100% MMAB exhibit low effective gate modulation, and 100% AB delivers insulating transport behaviour, closely resembling that of hBN (Supplementary Fig. 19). This result is also consistent with the low nucleation rate under 100% MMAB conditions (Fig. 1d) and the inspection of reaction free energies (Supplementary Fig. 2b), which indicate the desorption of $^*\text{H}_2\text{BNH}_2$ species and the adsorption of C-reactive species on the surface.

Moreover, we compared BCN FETs with emerging non-TMD p-type semiconductors. The comparison results are shown in Fig. 5f. The hole mobility of 2D BCN compares favourably to h- TiO_2 (10 nm) and β - Bi_2O_3 (the only two p-type metal oxides characterized by a high hole mobility over $100 \text{ cm}^2/\text{V-s}$ at room temperature). Particularly, 2D BCN surpasses these non-TMD p-type semiconductors in terms of scalability, I_{on} , V_{th} , and $I_{\text{on}}/I_{\text{off}}$, making it a viable candidate that meets the 2028 International Roadmap for Devices and Systems challenge for core logic devices. Next, to demonstrate wafer-scale uniformity across batch production, 224 BCN FETs from three growth batches were characterized at different positions across a 2-inch wafer. Three representative regions were selected (A, left; B, centre; C, bottom-left) as indicated in the optical wafer images. The statistical distributions of μ_{eff} , I_{on} , V_{th} and $I_{\text{on}}/I_{\text{off}}$ for all 224 devices are shown in Fig. 5g, in which the device index directly follows the spatial ordering from region A to B to C. The narrow spreads and the fact that most devices lie above the 2031 IRDS high-performance target (dashed lines) demonstrate uniform performance across the wafer. As the overall performance, epitaxial growth, scalability and uniformity are compatible with the current n-type MoS_2 standard, combining p-type BCN with n-type MoS_2 (ref. 28) is expected to produce high-performance, low-power complementary electronics, atomically thin circuits and junction devices. Apart from the continuous uniformity on flat metallic substrates, BCN can be deposited onto 3D-patterned insulating substrates in a conformal fashion. As demonstrated in Supplementary Fig. 20, monolayer BCN thin films can be selectively deposited on a holey SiO_2 dielectric rather than on a Si substrate. The compatibility with various topographies, in tandem with the selectivity towards amorphous dielectrics, could deliver a jolt to the scalable adoption of complex device architectures such as gate-all-around FETs (GAAFETs), vertical FETs and forksheet FETs, thereby paving the way for future commercialization.

In conclusion, we have demonstrated high-performance, wafer-scale p-channel FET arrays based on BCN semiconductors, enabled by precursor-engineered CVD synthesis. Through integrated multi-scale characterization—from atomic coordination and bonding configurations to nanoscale domain structures—we established a robust feedback loop between simulation, synthesis and device optimization. The resulting devices exhibit superior electrical performance compared with state-of-the-art p-type semiconductors, including 2D TMDs, non-TMDs, oxides and black phosphorus, while offering intrinsic stability, uniformity and industry compatibility. We thus predict that BCN can be a pivotal material to complete the p-type landscape.

Online content

Any methods, additional references, Nature Portfolio reporting summaries, source data, extended data, supplementary information, acknowledgements, peer review information; details of author contributions and competing interests; and statements of data and code availability are available at <https://doi.org/10.1038/s41586-026-11047-9>.

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Methods

BCN thin-film growth and characterizations

Monolayer BCN films were synthesized by low-pressure chemical vapour deposition in a 3-inch furnace tube equipped with two heating zones. The deposition process involved placing a 2-inch Cu (111)/sapphire substrate in the central heating zone of the main chamber. The precursors used were AB (97%, approximately 102 mg) and MMAB (99%, approximately 250 mg), which were loaded into two separate sub-chambers positioned upstream of the main chamber and modulated using mass flow controllers.

Initially, the furnace was evacuated to a base pressure of 5.0 torr. Before BCN growth, the substrate was annealed at 1,020 °C for 60 min under an H₂ flow of 200 standard cubic centimetres per minute (sccm). It is noted that humidity levels are typically around 75% in summer and 10% in winter. MMAB should be stored in the glovebox or vacuum desiccator. Therefore, the annealing temperature for precursors may vary. In general, the precursor sub-chambers were heated to 100 °C for AB and 130 °C for MMAB, respectively, using heating belts, and maintained at these temperatures for 10 min. Once the precursors reached the desired temperatures, they were introduced into the main chamber for 30 min to promote the growth of BCN on Cu (111)/sapphire substrates, which were oriented perpendicular to the diffusion of precursor vapours. After the growth process, the sub-chamber was sealed, and the main chamber was allowed to cool naturally to room temperature under a hydrogen gas flow of 50 sccm. The AFM images were obtained in the non-contact mode using a Park NX10 AFM instrument. A probe (AC160TS) with a force constant of 26 N m⁻¹ was used.

HR-STEM characterization

For the cross-sectional observation, thin TEM lamellas were prepared by an FEI Helios G4 FIB/SEM dual-beam system with a Ga-ion source, and the STEM images were taken by a JEOL NEOARM operating at 200 kV. To observe monolayer BCN in plain view, we acquired atomic-resolution HAADF-STEM images using an ARM300CF (JEOL) microscope. To minimize electron beam damage, the microscope was operated at 60 kV, and the beam current was about 20 pA.

STS/M characterization

STS/M measurements were performed using a commercial Omicron VT-STM system operating under ultrahigh vacuum conditions with a base pressure of 2×10^{-10} torr. Before spectroscopic analysis, the synthetic BCN film was thermally annealed at 500 °C for 6 h within the ultrahigh vacuum chamber to ensure surface cleanliness. All STS/M measurements were conducted at room temperature using constant-height mode with electrochemically etched tungsten tips prepared in NaOH solution to achieve atomically sharp probe geometries. The electronic band edges were determined from the onset positions in the differential conductance (dI/dV) spectra, which are proportional to the DOS of BCN. At the onset position, the tunnelling current begins to increase, indicating a transition from the bandgap region into accessible electronic states within the conduction or valence bands. This onset behaviour reflects the fundamental relationship between tunnelling probability and the availability of electronic states at specific energies, at which the sharp rise in differential conductance directly corresponds to the band edge energy.

DFT calculation

Vienna Ab Initio Simulation Package (VASP) was used to carry out the first-principles calculations based on DFT combined with the projector-augmented wave pseudopotential method. The Perdew–Burke–Ernzerhof functional within the generalized gradient approximation was selected to describe the exchange–correlation interaction. The cut-off energy was set at 450 eV for the plane-wave basis set.

For the decomposition pathway calculation of the AB and MMAB, the Cu (111) surface modelled by a 6×6 supercell containing six atomic layers was used. The Brillouin zone was sampled with $2 \times 2 \times 1$ Γ -centred k -point grids for the Cu supercell. A 10×10 Cu (111) supercell with $1 \times 1 \times 1$ k -point grid was chosen for the BCN nucleation and growth calculations. All the vacuum space along the surface normal is larger than 18 Å. DFT-D3 vdW correction was used to conduct the structure relaxation calculation. All geometric structures were relaxed until the electronic energy was below 1.0×10^{-5} eV. The special quasi-random structure with 72 atoms was built for the BCN monolayer using the Alloy Theoretic Automated Toolkit. The lattice constant of the BCN is 2.509 Å, which is extracted from the HR-STEM image. The DOS was calculated using the screened hybrid functional, HSE06, of Heyd, Ernzerhof and Scuseria.

Device fabrication and characterization

A heavily doped Si wafer (resistivity 0.001–0.009 Ω cm) with a 20-nm-thick ALD-grown HfO₂ was used as the gate electrode and the dielectric layer. The BCN monolayer channels were subjected to dry transfer onto HfO₂ as channel layers. Ni/Au source and drain electrodes were first defined by ultraviolet lithography (TuoTuo Technology) and then deposited by thermal evaporation to complete the bottom-gate top-contact FETs. The channel length and width were 1 μ m and 5 μ m, unless stated otherwise. All FET electrical characterizations were performed using a Keysight B1500A semiconductor analyser in ambient conditions. Mobilities were independently extracted from the linear-regime transconductance of both the forward and reverse I_d – V_g sweeps at $V_{ds} = -1$ V, using the standard expression. We note that the threshold-voltage shift between forward and reverse sweeps is very small ($\Delta V_{th} \approx 35$ mV), and therefore, the extracted mobilities from the two directions nearly overlap over the whole gate-voltage range. This demonstrates that hysteresis effects are minimal in our devices and that the reported high mobility values are robust with respect to the sweep direction (the forward- and reverse-extracted mobilities differ by less than 4% in the plateau region, confirming the reliability of the quoted average mobility).

Data availability

The data that support the findings of this study can be found in the paper and the Supplementary Information and are available from the corresponding authors upon request.

Code availability

The DFT calculations were performed using VASP v.6.3.2, which is commercially available at <https://www.vasp.at/>. The custom scripts and VASP input files generated during the current study are available from the corresponding authors upon request.

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Author contributions C.-C.T., J.-H.F. and V.T. conceived the study. C.-C.T., C.-H.H. and S.R.C. performed the experiments and analysed the data. J.-C.K., P.-Y.Y., Y.-C.L. and C.-W.P. performed the DFT and PES calculations. A.L., R.I. and N.S. performed the HR-STEM characterizations. K.Q. and C.-C.T. conducted the KPFM and STS/M. C.L. established the CVD setup. W.-C.C., Y.-W.L., C.-T.P., M.-C.W. and F.-Z.C. deposited the HfO₂ dielectric. K.N., Y.-L.H. and C.-Z.D. designed the electrode patterns and analysed the transfer and output characteristics of BCN FETs. C.-H.H., J.-C.K., Y.-C.C., J.-H.F., C.-W.P. and V.T. wrote the paper. All the authors discussed the results and contributed to the final version of this manuscript.

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Additional information

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