1 Self-assembled InN Micro-Mushrooms By 2 Upside-down Pendoeepitaxy

3 *ATM Golam Sarwar¹, Fan Yang², Bryan D Esser², Thomas F Kent², David W McComb², and*
4 *Roberto C Myers^{1,2*}*

5 ¹Department of Electrical and Computer Engineering, The Ohio State University, Columbus,
6 Ohio 43210, USA.

7 ²Department of Materials Science and Engineering, The Ohio State University, Columbus, Ohio
8 43210, USA

9 *E-mail: myers.1079@osu.edu

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ABSTRACT

Self-assembly of hexagonal InN micro-mushrooms on Si (111) substrates by molecular beam epitaxy is reported. Scanning electron microscopy (SEM) reveals hexagonal mushroom caps with smooth top surfaces and a step-like morphology at the bottom surface. A detailed growth study along with SEM measurements reveals that an upside-down pendeoepitaxy mechanism underlies the formation of these structures. Cryogenic temperature photoluminescence measurements on the InN disks show dominant band-to-acceptor recombination peak at 0.68 eV. Cross-section annular bright field (ABF-) scanning transmission electron microscopy (STEM) reveals that the growth of these structures occurs along the $[000\bar{1}]$ crystallographic orientation (N-face). Plan-view scanning transmission electron microscopy (HAADF-STEM) in the center of the micro disks reveals a hexagonal lattice indicative of stacking faults. However, at the outskirts of the micro disk, surprisingly, a honeycomb lattice is observed in plan-view STEM indicating a perfect free-standing Wurtzite InN disk that is free of stacking faults. This result opens a pathway for realizing strain-free, freestanding InN substrates.

KEYWORDS: B1 nitrides; A1 nanostructures; A3 pendeoepitaxy; A1 Planar defects; B2 Semiconducting indium compounds

INTRODUCTION

Advancement in epitaxial growth techniques has led to the discovery of new fundamental physics such as the quantum Hall effect [1], as well as game-changing technologies such as high electron mobility transistors (HEMTs) [2] and LEDs [3]. Though very high quality epitaxial films have been demonstrated for lattice matched or small lattice mismatched systems, acquiring high quality epitaxial films for materials without lattice matched substrates is still challenging. The bottleneck is strain induced relaxation by defect formation [4], [5]. The creation of misfit dislocations at the interface between the epitaxial film and the substrate is accompanied by threading dislocations extending to the surface [5]. These threading dislocations propagate in the epitaxial film from the substrate which degrade the performance of electronic and optoelectronic devices [6], [7].

Nanowires, on the other hand, are promising for realizing hetero-epitaxial layers with large lattice mismatch, as strain can relax via the nanowire sidewalls without the need for misfit dislocation formation. Also, threading dislocations do not propagate to the top surface, but terminate at the sidewalls near the base [8], [9], leaving the top part of the nanowire dislocation-free. As a result, extreme lattice mismatched heterostructures can be formed in nanowires which would not be possible in conventional planar films [10]–[19]. Nanowires also add another degree of freedom, as they can be grown on a variety of substrates including single-crystalline, polycrystalline, and amorphous materials [20]–[23]. Once high quality material is achieved, pendeoepitaxial overgrowth (a technique in which vertical and lateral growth occurs at the same time) can be used to grow high quality large area single crystals or low dislocation density thin films from nanowires or selectively grown columns [24]–[27].

One example of a material that lacks the availability of a lattice matched substrate is InN. For epitaxial growth, the lattice mismatch to GaN and AlN is ~11% and ~14%, respectively. As a result, the growth of InN on these substrates leads to highly defective material [28]–[31]. InN nanowire growth has been explored thoroughly on Si substrates over the last decade and high quality nearly defect free InN nanowires have been reported [32]–[35].

Here, we report the growth of hexagonal InN disks on nanowires on Si using a two-step growth method. The growth of GaN and InGaN disks on nanowires was previously observed using low temperature growth [27], [36], [37]. However, to the best of our knowledge, such structures are not previously reported for the case of InN. Unlike GaN and InGaN, the InN disks form at high temperatures compared to the growth temperature of InN nanowires. In our study, the InN nanowires are nucleated at relatively low temperature, followed by growth at relatively high temperature. This results in InN micro-disks on top of InN nanowires resembling mushroom-like structures. Additional growth studies combined with scanning electron microscopy (SEM) observations reveal an upside-down pendeoepitaxy mechanism that is responsible for the formation of the micro-mushrooms. Low temperature micro-photoluminescence of an individual InN micro-disk shows dominant InN band-to-acceptor carrier recombination. Cross-sectional and plan-view high angle annular dark field (HAADF) and annular bright field (ABF) scanning transmission electron microscopy (STEM) imaging is carried out revealing a perfect Wurtzite InN lattice at the outskirts of the mushrooms, which is free of stacking faults.

RESULTS AND DISCUSSIONS

We grow InN nanowires on Si using plasma assisted molecular beam epitaxy. All the samples grown for this study are summarized in Table 1. First, InN nanowires are nucleated for 15

76 minutes at 400°C followed by a 75 minute growth at 450°C (**Sample A**). Figure 1 shows the
77 SEM images of InN grown using the described two-step growth process. Shown in Figure 1a, we
78 observe a hexagonal InN micro-disk on top of an InN nanowire, or micro-mushroom, on Si. The
79 diameter of the hexagonal cap (disk) of the mushroom ranges from less than 1 μm to ~3 μm,
80 while the average diameter of the stem is ~250 nm. In the plan-view image (Figure 1b) the stem
81 of the mushroom is visible, suggesting that the top hexagonal InN disk is electron transparent.
82 Figure 1c shows the tilted view SEM image of the mushrooms where the flat hexagonal top
83 surface can be observed. To observe the bottom surface of the hexagonal InN cap, we sonicate
84 the mushrooms in an isopropanol solution and transfer them to a SiO₂ substrate. Figure 1d shows
85 an SEM image of a flipped InN cap after the transfer process. The bottom surface of the InN cap
86 shows step-like morphologies. The bright spot at the center of the cap represents the broken InN
87 stem (nanowire). The steps at the bottom surface can clearly be observed in a magnified SEM
88 image shown in Figure 1e. The average lateral step size is measured to be ~15 nm.

89 We carried out additional experiments to understand the growth mechanism of these interesting
90 structures. **Sample B**: InN nanowires are grown for 90 minutes on Si at 400°C. **Sample C**: InN
91 nanowires are first grown in the same fashion as sample B, then the substrate temperature is
92 raised and kept at 450°C for 15 minutes. During the 450°C step, the N-plasma was on while the
93 In shutter was closed. By doing so, we ensure a supply of active nitrogen on the growth surface
94 without any active In from the source. Figure 2a shows the SEM image of sample B. Nearly
95 vertically aligned InN nanowires are observed on the Si substrate. Though tapering of InN
96 nanowires is observed, the formation of mushrooms is not. Figure 2b shows the tilted view SEM
97 image of sample C. Hexagonal InN disks are observed on the growth surface of sample C. Figure
98 2c shows a magnified SEM image of an InN disk. Hexagonal InN disk with flat top surface

similar to the cap of the InN micro-mushroom (Figure 1) is observed. The SEM image of a flipped hexagonal InN disk on a transferred surface (Figure 2d) shows step-like morphologies at the bottom surface of the disk, which is also observed at the bottom surface of InN micro-mushroom caps (Figure 1e). From the comparison of the SEM images of sample B and C, it is evident that the InN disks (in sample C) form during the high temperature (450°C) N-plasma anneal step.

The formation of the disks at high growth temperature (450°C) without the supply of In from the source (sample C) demonstrates that the In incorporated into the lateral caps must come from the bottom of the nanowires. Previous studies on GaN and InGaN revealed that the lateral overgrowth i.e. disk formation occurs at low temperatures compare to the growth temperatures of nanowires [27], [36], [37]. In contrast, here, InN disks form at high temperature compared to regular InN nanowire growth temperature, which can be explained from the fundamental difference in decomposition and desorption temperature between InN and GaN. It is well established that the decomposition temperature of InN is much lower than the desorption temperature of In. At elevated temperature (450°C), upon InN decomposition, N atoms can escape the InN surface by forming N₂ leaving In adatoms on the surface. These In adatoms diffuse along the surface of the nanowires and in the presence of active N-plasma reincorporate into InN. Due to the high adatom mobility of these atoms at elevated temperature, a large portion of the decomposed In is expected to reach the growth front (top of the nanowire) and incorporate there (vertical growth). The increased concentration of In adatoms at the growth front would increase the local III/V ratio (for GaN, the increase in local III/V ratio thus increased lateral overgrowth [27], [36], [37] occurs at low temperature because of lower Ga desorption temperature compared to GaN decomposition temperature) and drive the lateral growth to form

InN disk, thus named upside-down pendeoepitaxy. We speculate that this whole process occurs in a layer by layer fashion and causes the step-like morphology at the bottom surface of the InN mushroom caps or disks. Due to decomposition, the nanowire base gets smaller or completely disappears (Figure 2c and d). However, when In supply from the source is on (sample A), In diffusion from the Si substrate, to some extent, balances the InN base decomposition and the nanowire base remains (as in Figure 1). The described growth mechanism of InN micro-mushrooms is schematically shown in Figure 2e. The InN nanostructures of this study grow in the $[000\bar{1}]$ crystal orientation or N-face (shown later). The decomposition temperature reported for metal rich N-face planar InN growth is much higher compared to the reported values in this study [38]. It is important to note here that all substrate temperatures reported in this study are thermocouple (TC) readings. The substrate temperature can be measured using a thermocouple or a pyrometer. Figure S1 (in supporting information) shows the available pyrometer readings vs thermocouple measurements. The pyrometer readings are $30^{\circ}\text{C}\sim 50^{\circ}\text{C}$ higher than the thermocouple setpoints. Because the pyrometer does not read temperature values over the entire range of interest ($\text{TC} = 400^{\circ}\text{C} - 450^{\circ}\text{C}$) and at its limit of detection, in this study, we do not present our pyrometer readings.

Next, we probe a single hexagonal InN disk transferred on a SiO_2 substrate with micro-photoluminescence ($\mu\text{-PL}$) cooled to cryogenic temperature (11 K). Figure 3a shows the $\mu\text{-PL}$ spectrum of a single InN disk with increasing excitation power with the top right inset showing semi-logarithmic plot. The small peak at ~ 0.77 eV (1600 nm) is the 2nd order diffraction from the laser. Apart from that, we observed three different characteristic peaks in the PL spectrum. The strongest PL emission is observed at ~ 0.68 eV. We also observe two sideband weak emission peaks: one at high energy (~ 0.72 eV) and one at low energy (~ 0.62 eV). The dominant emission

peaking at ~ 0.68 eV is in good agreement with previously reported PL emission from nearly intrinsic InN nanowires [39], [40] and InN thin films with low background electron concentration [41]–[44]. The integrated PL (bottom right inset of Figure 3a) shows a linear increase with increasing optical injection power. A full width at half maximum (FWHM) of ~ 27 meV is observed for an injected optical power of 2.32 kW/cm^2 . The PL FWHM of state-of-the-art InN nanowires is reported < 10 meV at 10K with an excitation power of $\sim 4 \text{ } \mu\text{W}$ [45]. It is also reported to increase from ~ 15 meV to ~ 40 meV with the increase of excitation power from $5 \text{ } \mu\text{W}$ ($\sim 0.025 \text{ kW/cm}^2$) to 9 mW ($\sim 46 \text{ kW/cm}^2$) at 20 K [46]. So, the PL FWHM observed from InN disk in this study is comparable to the PL FWHM observed for InN nanowires. Such a narrow FWHM (at sufficiently large injection power) and dominating peak at ~ 0.68 eV also suggest the presence of small background electron concentration in these InN disks [46]. Furthermore, we observe a blue shift of the peak emission energy and increase of FWHM upon increasing the excitation intensity. Figure 3b shows peak emission energy (red squares), and FWHM (blue circles) as a function of laser power. Peak emission energy and FWHM increases from 683.7 to 688.2 meV, and 27.3 to 39.2 meV, respectively as the laser power increases from 2.32 to 29.16 kW/cm^2 and can be attributed to the band filling effect due to photogenerated carriers. These observations (linear increase of integrated PL, increase in FWHM, and blue shift of the peak energy) suggest that band-to-acceptor radiative transition is the mechanism responsible for the dominant PL emission [41], [43], [44].

The weak low energy PL peak (~ 0.62 eV) has previously been observed in both thin films and nanowires, and is attributed to either band-to-deep acceptor transition [41], [42] or phonon side band emission [39], [47]. We also observe a weak high energy peak at ~ 0.72 eV. The co-existence of band-to-acceptor PL (~ 0.68 eV) with a dominant high energy peak (~ 0.72 eV) was

previously observed in InN thin films with high background electron concentration [42]. It is worth mentioning here that the PL from InN nanowires exhibit the ~ 0.68 eV band-to-acceptor emission peak with no high-energy sideband. For comparison, normalized PL emissions of InN nanowires and disks are shown in Figure 3c. While the observation of dominant ~ 0.68 eV PL peak suggests low background electron concentration, the high energy peak at ~ 0.72 eV indicates the presence of high electron concentration [39], [41], [42], [44], [48]. We explain this phenomenon by surface electron accumulation in the top surface of the InN disks, shown schematically in Figure 3d. We think the top surface of the InN disk suffers from high electron accumulation due to surface fermi pinning at the c-plane while the bulk of the disk has low background electron concentration. Thus, two recombination mechanisms can be active in these InN disks. (1) Recombination away from the surface (or bulk; E_1 in Figure 3d): electrons at the bottom edge of the conduction band recombine with localized holes in the acceptor level, and (2) Recombination near the surface (E_2 in Figure 3d): electrons near the Fermi energy recombine with localized holes in the acceptor level. We attribute ~ 0.68 eV and ~ 0.72 eV peaks to processes E_1 and E_2 , respectively (Figure 3d).

The interesting growth mechanism and superior photoluminescence characteristics of these micro-structures led us to perform structural characterization using advanced microscopy techniques. The growth orientation of III-N Wurtzite materials can be identified using cross-sectional ABF-STEM imaging where both heavy (In) and light (N) atoms can be observed easily. Figure 4a shows the cross-sectional ABF-STEM image of an InN micro disk. Fitting the ABF-STEM image with cross-section Wurtzite lattice pattern reveals that InN nanowires grow in the $[000\bar{1}]$ orientation or N-face. Figure 4b shows the experimentally averaged ABF-STEM image of an InN repeat unit in the $[1\bar{1}00]$ direction. The low intensity sites (black) represent In sites,

while intermediate intensity sites (gray) represent N sites. The simulated repeat cell is also shown (Figure 4c) for visual convenience.

Cross-sectional STEM imaging of semiconductor epitaxial layers is commonly performed because of relatively easy sample preparation techniques. In contrast, plan-view STEM imaging studies of semiconductor materials are rarely done due to very complex sample preparation methods [49]. The electron transparent nature of the thin InN disks (observed in plan-view SEM image of Figure 1b) allows us to study plan-view STEM without any need for complex sample preparation steps.

Figure 5a and b show a plan-view (c-plane) STEM image (from sample C) and schematic of an InN disk, respectively. The thickness variation and the step-like morphology are clearly observed in Figure 5a. Figure 5c shows plan-view HAADF-STEM image from the central area of the InN disk as indicated in the schematic diagram of Figure 5b. We observe hexagonal lattice in this area (Figure 5c). The inset of Figure 5c shows a magnified image with a schematic of the hexagonal lattice. More examples of such observations are given in the supporting information (Figure S2). In Figure 5b, we also show the ideal plan-view (bottom left) and cross-section (bottom right) of the Wurtzite crystal structure. The atomic columns observed from the top (c-plane) are all equivalent (dashed lines in the bottom right of Figure 5b) and a honeycomb lattice, unlike Figure 5c where hexagonal lattice is observed, with equal Z-contrast of each lattice site is expected. This discrepancy is explained by the projection of stacking faults into the plan-view images; stacking faults are observed in cross-sectional STEM images discussed below.

Figure 6a shows a cross-sectional STEM image taken near the top surface of the InN disk. We observe only Wurtzite stacking near the top surface. The inset shows the FFT of the HAADF-

213 STEM image and mimics the expected FFT pattern for a Wurtzite crystal. As we move away
 214 from the top surface we observe a large number of stacking faults (Figure 6b and S5). Both
 215 intrinsic (I1 and I2) and extrinsic (E) stacking faults are observed in the STEM image analysis
 216 (Figure S5 and S6). We also observe thick (up to 11 bi-atomic layer) zinc blende insertions
 217 (Figure 6b and S6). The inset of Figure 6b shows the FFT of the yellow dashed box of the STEM
 218 image and resembles the zinc blende FFT pattern. The presence of stacking faults in cross-
 219 sectional STEM images can be linked to the observation of hexagonal (instead of ideally
 220 expected honeycomb) lattice pattern in plan-view. In an ideal Wurtzite lattice (**A-B** stacking and
 221 **C** absent – see Figure S7), a honeycomb lattice is expected in the plan-view STEM image. Here,
 222 in the relatively thicker central region of the disk, due to the stacking faults, the **C** stacking
 223 appears (Figure S7). As the plan-view STEM image is a projection of the atomic arrangement on
 224 the c-plane over the total thickness of the disk, we observe hexagonal (Figure 5c) lattice. Unlike
 225 the ideal honeycomb lattice (Figure 5b), where all the atomic columns are equivalent and
 226 expected to give the same HAADF-STEM intensity, the hexagonal lattice (Figure 5c) gives
 227 different HAADF-STEM intensity on different lattice sites (**A**, **B**, and **C**). For example, if the
 228 atomic stacking switches from **A-B** to **A-C** due to stacking fault, **A** lattice sites are expected to
 229 give higher HAADF-STEM intensity than **B** and **C** lattice sites. This variation of intensity is
 230 observed in the plan-view HAADF-STEM images when the hexagonal lattice is observed
 231 (Figure 5c and S2). Using the quantum excitation of phonons model [50], we perform a
 232 simulation of the plan-view STEM image of Wurtzite InN with the insertion of stacking faults.
 233 Figure 5d shows the simulated STEM image with 10 pairs **A-B**, 8 pairs **C-A**, and 7 pairs **B-C**
 234 stacking (Figure S8), which corresponds to 18, 17, and 15 atoms in **A**, **B**, and **C** sites,
 235 respectively. The intensity variation from site to site is easily distinguishable for even such small

variations in the number of atoms in different sites (18, 17, and 15). The simulated intensity variation is much more prominent where differences in the numbers of atoms are larger (Figure S9).

However, as we move toward the outskirts of the disk (relatively thin region) we observe the disappearance of the hexagonal lattice and appearance of a honeycomb lattice. Figure 5e shows plan-view HAADF-STEM images from the outskirts of InN disk as indicated in the schematic diagram of Figure 5b. This thinnest section in the plan-view represents the top surface of the cross-sectional view of InN disk and found to be free of stacking faults (Figure 6a). As a result, a honeycomb lattice with equal Z-contrast of each lattice site is observed in this region as expected in a plan-view of ideal Wurtzite crystal (bottom left of Figure 5b). The bottom left inset of Figure 5e shows a magnified section with a schematic of the honeycomb lattice (yellow circles). A 2D unit cell of the Wurtzite structure is also shown (red circles). An extracted in-plane lattice constant of ~ 3.5 Å is in close agreement with the lattice constant of InN. The top right inset shows an FFT of the HAADF-STEM image. Figure 5f shows a simulated plan-view (c-plane) STEM image of an ideal Wurtzite InN crystal confirming the honeycomb lattice with equal intensity in each lattice point. More plan-view STEM images of the honeycomb lattice can be found in the supporting information (Figure S3). This is, to the best of our knowledge, the clearest observation of the c-plane honeycomb lattice in any Wurtzite material system [49], [51]–[54]. In some cases we also clearly observe the transition from honeycomb to hexagonal lattice (Figure S4).

The observation of a honeycomb lattice in the plan-view STEM indicates the absence of stacking faults within the free-standing outskirts of the disks. The InN disk on nanowire or micro-mushroom can therefore be investigated for use in realizing strain-free, freestanding InN

substrates by coalescence of such lateral structures. Following previous work along these lines in GaN and InGaN, InN nanowires can be grown on patterned Wurtzite substrates followed by the growth of hexagonal InN disks on nanowires exploiting upside-down pendeoepitaxy. The lateral growth process can be continued until the hexagonal disks merge together to form large area strain-free InN thin film on top of the InN nanowires. Once complete merging of the disks is achieved freestanding InN substrates can be realized by growing for longer amount of time. Currently, the growth of high quality InN thin films is hindered by the scarcity of lattice matched substrates. The upside-down pendeoepitaxy described in this work can open a viable pathway toward the formation of strain-free, freestanding high quality InN without the need for lattice matched substrates.

CONCLUSIONS

A new growth mechanism, called upside-down pendeoepitaxy is described which leads to the formation of InN micro-mushrooms by molecular beam epitaxy. Micro-photoluminescence measurements show dominant band-to-acceptor recombination at 0.68 eV with two emission peaks at 0.62 eV and 0.72 eV. The high-energy peak (0.72 eV) is attributed to a band-to-acceptor transition from the electron accumulation layer on the top surface, while the low energy peak may be due to a band-to-deep-acceptor transition or phonon side band emission. The growth direction of the InN nanostructures on Si is determined to be $[000\bar{1}]$ (N-face) from cross-sectional ABF-STEM. The electron transparent nature of the InN disk allows us to study them in plan-view using advanced STEM. The plan-view HAADF-STEM of the center of the InN disk reveals hexagonal lattice resulting from a series of stacking faults observed in the cross-sectional view. However, an ideal honeycomb lattice and Wurtzite lattice stacking is clearly observed at the outskirts of the free-standing InN disks in the plan-view and cross-sectional view STEM

images, respectively. The achievement of stacking fault-free InN disk opens a pathway toward the realization of strain-free, freestanding InN substrates.

EXPERIMENTAL SECTION

Material Synthesis. Self-assembled catalyst-free InN nanowires are grown on Si (111) using plasma assisted molecular beam epitaxy in a Veeco GEN 930 system with an average base pressure of $\sim 7 \times 10^{-11}$ Torr. The Si wafers are cleaned using a standard solvent cleaning process followed by a 200°C load lock chamber bake (4 hrs) and 800°C buffer chamber bake (1 hr) before transferring to the growth chamber. The native oxide from the Si wafer is desorbed by heating the substrate *in-situ* to 1100°C. The substrate is then cooled down to growth temperature. The In flux is measured to be 3.9×10^{-8} torr using a beam flux monitor (BFM) ion gauge which corresponds to a III/V ratio of 0.16. The N-plasma is lit with an N₂ partial pressure of 2×10^{-5} torr and plasma power of 350 W. The stoichiometric In flux is estimated to be $\sim 2.5 \times 10^{-7}$ torr, which gives an In rich growth rate of ~ 300 nm/hr.

Structural Characterization. An FEI/Philips Sirion SEM with a field emission (FE) source and an in-lens secondary electron detector was used for collecting SEM images. STEM imaging of the InN micro-structures were collected using an FEI image corrected Titan³ G2 60-300 S/TEM operated at 300 kV. To prepare plan-view STEM samples, a TEM Cu grid with an amorphous carbon support film was swabbed on sample C to pick up the floating InN disks. Cross-sectional STEM thin foil samples were prepared using an FEI Helios NanoLab 600 dual beam focused ion beam (FIB) instrument.

Optical Characterization. The InN disk was optically excited using a mode-locked Ti:sapphire oscillator (Coherent Chameleon Ultra II) operating at 800 nm and 40 MHz. The sample was

304 illuminated through a 0.5 NA 36× reflective objective **which results in 10 μm spot size** and the
305 emission was collected through a 1300 nm long pass filter and passed to a 0.5 m spectrometer
306 (Princeton Instruments SP2500i) equipped with a liquid nitrogen cooled InGaAs diode-array
307 sensitive from 1100 to 2200 nm (OMAV-2.2).

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465 **Table 1.** Summary of the samples

Sample name	Growth process
Sample A: InN micro-mushroom	15 minutes of InN nanowire nucleation at 400°C followed by 75 minutes of InN grown at 450°C
Sample B: InN nanowire	90 minutes of InN nanowire at 400°C
Sample C: InN micro-disk	90 minutes of InN nanowire at 400°C followed by 15 minutes of 450°C high temperature anneal step in N-plasma in absence of In flux.

466

467

Figure Captions:

Figure 1. Scanning electron microscopy (SEM) images of InN micro-mushrooms on Si. (a) Cross-sectional, (b) plan-view, and (c) tilted-view SEM images of InN micro-mushrooms showing hexagonal InN micro disk (cap of the mushroom) on top of InN nanowire (stem of the mushroom). The stem of the mushroom can be observed in the plan view SEM images which indicates that the top InN disk is electron transparent. (d) Bottom view SEM image of an InN disk transferred on a foreign substrate. (e) Magnified SEM image of dashed box of (d) showing the step-like morphology at the bottom of a InN disk. The scale bar is 200 nm.

Figure 2. Growth method of InN micro-mushrooms. (a) SEM image of InN nanowires using single step growth at 400 °C for 90 minutes. (b) SEM image of InN nanowires grown at 400 °C for 90 minutes followed by a 15 minutes N-plasma annealing at 450 °C. (c) Tilted-view SEM image of hexagonal InN disk formed during the N-plasma anneal. (d) Bottom-view SEM image of an InN disk transferred on a foreign substrate. (e) Schematic of the proposed formation mechanism of InN micro-mushrooms.

Figure 3. Cryogenic temperature micro-photoluminescence (μ -PL) of a single InN disk. (a) μ -PL spectrum as a function of excitation power. Insets show PL in semilogarithmic scale (top right) and integrated PL as a function of excitation power (bottom right). (b) Peak energy (red squares), and FWHM (blue circles) as a function of incident power. Lines are drawn to guide eyes. (c) Normalized PL spectra of InN nanowires and disk. (d) Schematic of surface band bending and electron accumulation. Schematic of different electron to acceptor transitions are also shown.

Figure 4. Cross-sectional annular bright field scanning transmission electron microscopy (ABF-STEM) characterization of InN disk. (a) Cross-sectional ABF-STEM image showing relatively

light N-atoms. Top left corner shows schematic of InN crystal (green: In; blue: N). Experimentally averaged (b) and simulated (c) ABF-STEM image of an InN repeat unit along $[1\bar{1}00]$ direction. Low intensity sites (black) are In, while intermediate intensity sites (gray) are N.

Figure 5. Plan-view high angle annular dark field scanning transmission electron microscopy (HAADF-STEM) characterization of InN disk. (a) Plan-view HAADF-STEM image of an InN disk showing step-like morphologies. (b) Schematic of a InN disk (top), ideal plan-view (bottom left), and ideal cross-section (bottom right) of atomic columns. (c) Plan-view HAADF-STEM image taken at the central area of the InN disk (indicated in Figure 3(b)). Hexagonal lattice is observed. Inset shows magnified atomic resolution image. (d) Plan-view HAADF-STEM image simulation for InN with stacking faults. (e) Plan-view HAADF-STEM image showing honeycomb lattice at the outskirt (indicated in Figure 3(b)) of the InN disk. Insets show magnified image with 2D unit cell (bottom left) and FFT of the HAADF-STEM image (top right). (f) Plan-view HAADF-STEM image simulation for InN without stacking faults. Scale bars are 1 nm if not mentioned otherwise.

Figure 6. Cross-sectional high angle annular dark field scanning transmission electron microscopy (HAADF-STEM) characterization of InN disk. (a) Cross-sectional HAADF-STEM image taken near the top of the InN disk showing only Wurtzite stacking. (b) Large number of stacking faults observed (shown using arrows and curly brackets) in a cross-sectional HAADF-STEM image taken away from the top surface. Scale bars are 1 nm.

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