Synthesis, Characterizations and Applications of Cadmium Chalcogenide Nanowires: A Review

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Cadmium chalcogenide nanowires have demonstrated superior electrical and optical properties, and have emerged as prominent building blocks for nanoscale electronic and optoelectronic devices. In addition to the effort devoted to advance techniques of fabricating high quality nanowires, much has been endeavored to elucidate their unique physical properties for better design and development of functional devices with low power consumption and high performance. Herein, this article provides a comprehensive review of the forefront research on cadmium chalcogenide nanowires, ranging from material synthesis, property characterizations, and device applications.

Key words: Cadmium chalcogenide; Nanowire synthesis; Electrical transport; Photoluminescence; Band diagram; Nanoelectronics; Optoelectronics

1. Introduction

Nanoscience and nanotechnology are multidisciplinary fields of research conducted at the nanoscale, on the order of 10^{-9} m. They not only offer great challenges in many of the fundamental aspects, but also drive technological innovations that exert significant impact, in electronics, photonics, spintronics, health care, to as far as space exploration. These interdisciplinary fields are where tiny elementary entities are fabricated and investigated, such as nanowires, nanotubes, nanoparticles, proteins, DNA, etc. The ability to synthesize these nanoscale modular units and then to assemble them into larger structures

with unique properties in charge conduction, optical guiding, and magnetic ordering, is changing the way to produce materials and devices.

Among various nanomaterials, structures with quasi-one-dimensionality (Q1D) have been a research focal point ever since the discovery of carbon nanotube in 1991^[1]. Due to the lack of efficient methods to fabricate carbon nanotubes with specific electrical characteristics, tremendous effort has been devoted to the study of semiconducting nanowires with high crystalline structures and controllable physical properties. In the current trend of device miniaturization, these Q1D nanostructures play an important role as the building blocks for devices that could overcome the fundamental limits of microtechnology. Most material properties deviate from bulk behaviors when the size of one dimension is reduced below a few hundred nanometers; and in nanowires this change is much more pronounced because two dimensions are dramatically reduced. To a certain extent, nanowires open a new system of materials since the properties get altered as a result of the dimensional confinements. To explore this new material system, a fundamental understanding is crucial. Only with a deeper comprehension of the basic characteristics of nanostructures, one can better tailor their properties for designing and developing innovative devices. In essence, the unique physical properties originate from the Q1D structures with highly anisotropic geometry and enlarged surface areas; furthermore, when the size approaches fundamental length scales, such as Bohr radius or localization length, quantum effects will reveal. Q1D structure possesses large polarization anisotropies in optical absorption and emission mainly due to the geometric anisotropy, leading to excellent polarization sensitive devices^[2]. Moreover, Q1D materials system has a much larger surface-to-volume ratio (inverse proportional to the radius) than its bulk counterpart, thus dramatically enhancing functionalities that base on surface sensitivity^[3]. On the other side, in the quantum confinement region, electronic band structure is strongly size-dependent, enabling a precise knob for fine-tuning the electrical, optical, and chemical properties^[4].

The device prototypes based on semiconductor nanowires have already demonstrated groundbreaking advances, as they have been utilized as chemical sensors^[3,5], field effect transistors^[3,6,7], photodetectors^[8–10], nanolasers^[11,12], solar cells^[13], etc. Among semiconductor Q1D materials, cadmium chalcogenide (CdS, CdSe and CdTe) nanowires have emerged to the research forefront because of its exceptional properties and broad device applications. They have direct band gaps that span from infrared to visible range, excellent light absorption capability, and superior photosensitivity. All of these make them outstanding materials for developing nanoelectronic, optoelectronic, and electromechanical

devices. To highlight the current research achievements and trigger more cutting edge investigations in this field, this article aims to provide a comprehensive summary of the research activities on the synthesis techniques, physical properties and device applications of the cadmium chalcogenides nanowires.

2. Synthesis

A variety of methods have been developed to synthesize cadmium chalcogenides nanowires. According to the synthesis environments, they can be mainly divided into two categories: solution phase growth and vapor phase growth. These methods will be discussed in details in this section. In addition to homogeneous nanowires, heterojunction nanowires are important for innovative functional nanodevices. Some prototypes of heterojunction nanowires proposed will also be presented in this section.

2.1. Solution phase method

Solution phase methods provide low temperature, cost-effective venues for cadmium chalcogenides nanowire growth. Various solution phase growth techniques have been successfully implemented. They include electrochemical deposition^[14–19], chemical bath deposition^[20], sol-gel method^[21], solvothermal^[22–31], sonothermal^[32], hydrothermal growth^[33, 34], etc. Together they can be classified into template directed and template-free growth methods.

2.1.1. Template directed growth

Anodic aluminum oxide (AAO) membrane that consists of hexagonally ordered nanopores has been the most commonly used template for fabricating aligned semiconductor nanowire arrays because of its chemical stability, insulating characteristics and good mechanical strength. More importantly, the morphologies of AAO (pore size, pore-to-pore distance and channel length) can be readily adjusted via anodization parameters, providing a straightforward route to fabricate nanowires of various sizes, including those within the quantum confinement region (< 10 nm in diameter)^[14].

Cadmium chalcogenides nanowires have been fabricated by filling the pores of AAO by different deposition techniques such as electrochemical deposition^[14,18–21,35–40], sol-gel method^[21], chemical bath deposition^[20], etc. Among them, electrochemical deposition is the most important and widely used method, by which CdS^[14,18,20,21,35,36], CdSe^[37] and CdTe^[19,38–40] nanowires have been obtained. However,

AAO templates have intrinsic insulating barrier layers at the bottom, which limit direct DC deposition. This issue can be either overcome by firstly thinning the barrier layer to ~10 nm by gradually reducing anodization voltage and then performing AC deposition^[14,18]; or by removing the barrier layer via chemical etching and then evaporating electrode for DC deposition^[15–17]. Both techniques have advantages and disadvantages. In DC deposition, removing the barrier requires more complicated processes, but it provides a more stable and uniform deposition. In both techniques, deposition rate depends highly on pore size, deposition voltage and current, concentration and temperature of electrolytes. Therefore, selection of electrolytes and control of the deposition parameters are critical to ensure a successful fabrication.

Besides AAO templates, other templates have also been utilized for guiding nanowire growth, but they are not as efficient as AAO templates in terms of sizes and shapes regulation. For example, CdS nanowires have been fabricated by hexagonal liquid crystal templates^[41], micelle-template^[42], and polymer ion track template^[43,44]. CdSe nanowires have been synthesized by using DNA^[45] and polycarbonate membranes^[46]. And CdTe nanowires have been obtained by electrodepositing in polycarbonate membranes^[47].

2.1.2. Template free methods

Apart from template filling method, many groups have ventured to develop new template free techniques for cadmium chalcogenide nanowire synthesis in the solution phase environment. Among them, solution-liquid-solid (SLS) growth has proven to be the most successful method for growing narrow and high quality nanowires, and solvothermal method is also widely utilized due to cost-effectiveness and easy handling. In the discussion below, SLS and solvothermal growth will be focused, along with other template-free solution phase growth techniques.

SLS is a solution based catalyzed growth method, in which nanoscale metallic colloids are used as catalysts^[48]. As the colloids heat up, they form droplets and catalyze the decomposition of metal-organic precursors, and foster crystalline nanowire growth^[48]. Because low synthesis temperature is required for SLS growth in solution environment, metal catalysts are restricted to those with very low melting points, such as Bi, In, Sn and Ga. Sizes of nanowires can be effectively controlled by sizes of metal catalysts, and thus this method is feasible for producing nanowires of the smallest sizes (< 10 nm) under low growth temperature (200–300 °C). CdS^[49], CdSe^[50–57] and CdTe^[50,57,58] nanowires with narrow diameter

have been fabricated by SLS method, which opens up remarkable opportunities to study the quantum confinement effects in the cadmium chalcogenide nanowires.

Solvothermal growth offers a promising synthetic method because of its effectiveness, convenience, and less energy consuming characteristics. It can be used to fabricate well-crystallized inorganic nanomaterials with tunable sizes and morphologies by employing different reactants, solvents, surfactants, and coordination agents. Various precursors and solvents have been used for growing CdS^[22–31,59–61], CdSe^[62,63], and CdTe^[62] nanowires. For example, Wang et al. have fabricated CdS nanowires by utilizing CdCl₂ and S powder as source materials and ethylenediamine as solvent^[59]. Phuruangrat et al. have reported CdS nanowires synthesized by the solution containing Cd(NO₃)₂, (NH₄)₂S and ethylenediamine^[60]. And Xu et al. have demonstrated CdS nanowires solvothermally synthesized in a mixed solvent of ethylenediamine and dodecanethiol^[61]. Moreover, polymer assisted solvothermal growth routes are also proposed in order to improve the uniformity of the diameters. For example, polyacrylamide (PAA)^[64,65] and polyvinyl alcohol (PVA)^[62] are employed as the coordination agents to stabilize the ion concentrations, so to optimize the nanowire growth. In addition, ligand-assisted solvothermal growth has also been proposed to fabricate CdS nanowires with high aspect ratio up to 100^[66].

In addition to the above commonly-used SLS and solvothermal growth techniques, other template-free methods have also been developed for growing cadmium chalcogenides nanowires such as surfactant-assisted growth^[67], sonochemical synthesis^[32], hydrothermal growth^[33,34], and cluster reaction methods^[67]. Some indirect growth processes are also proposed. For instance, by cation-exchange method, CdSe nanowires are obtained from Ag₂Se nanowires via the Ag⁺/Cd²⁺ cation-exchange process^[68]. Also, CdTe nanowires have been synthesized by transforming ZnTe into CdTe through vapor phase reaction^[69].

2.2. Vapor phase growth

Vapor phase growth is a well-adopted method to fabricate nanowires, formed by condensation of vaporized sources. Various vapor phase growth techniques have been developed to fabricate cadmium chalcogenides nanowires, including chemical vapor deposition (CVD)^[70–79], metal-organic chemical vapor deposition (MOCVD)^[80,81], physical vapor deposition (PVD)^[82], pulse laser deposition (PLD)^[83],

molecular beam epitaxial (MBE)^[84], etc. In general, the vapor phase growth of nanowires is governed by two different growth mechanisms: vapor-liquid-solid (VLS) and vapor-solid (VS).

2.2.1. Vapor-liquid-solid mechanism

The vapor-liquid-solid (VLS) mechanism was first proposed in 1964 by Wagner and Ellis as an explanation for the Si whisker growth^[85], which was then already widely applied for various whisker material growth. Metal nanoparticles, typically Au, are used in the VLS processes, which act as catalysts that facilitate uni-directional material growth. The VLS growth consists of several processes: A metal catalyst first absorbs source vapor and forms a eutectic alloy in the form of a liquid droplet; The liquid droplet has a much larger sticking coefficient and thus serves as a preferred site for deposition from the source vapor; Upon super-saturation, nucleation starts and drives the source to precipitate at the liquid alloy-solid interface and triggers the axial growth. The growth involves subsequently vapor, liquid and solid phases, thus is referred to as the VLS growth mechanism. It is essential to have eutectic alloy droplets in such VLS processes; therefore, equilibrium phase diagrams are essential for selecting optimal catalysts and growth conditions. Morphologies of nanowires grown by VLS depend upon sizes and physical properties of the liquid alloy. The resulted nanowires exhibit high crystalline qualities because of epitaxial growth. VLS growth has been successfully employed to fabricate single crystalline CdS^[70–74,83,86], CdSe^[75] and CdTe nanowires^[87–91]. **Fig. 1** shows a typical experimental set-up, schematic diagram showing the key processes in Sn-catalyzed VLS synthesis of CdS nanowires, accompanied by the corresponding phase diagram. Diameters of nanowires grown via VLS are usually larger than 10 nm, larger than that synthesized from SLS method mentioned in section 2.1.2. The difference lies in the growth temperature. At higher growth temperature (above the eutectic temperature) adopted in VLS, higher solubility in metal catalysts results in enlarged droplets during the alloying process. On the other hand, VLS growth can also occur without additional metal nanoparticles, via the so-called self-catalyst VLS growth. For example, CdS nanowires have been fabricated in a self-catalytic method, where S deficit CdS_{1-x} liquid droplets serve as catalysts^[92].

Furthermore, AAO template directed VLS growth also attracted a great research attention since it is advantageous to fabricate vertical aligned nanowire arrays with high material quality and controllable geometries^[93]. Metal nanoparticles are first deposited into the AAO channels by electrochemical deposition, and they subsequently catalyze nanowire growth along the cylindrical channels in a vapor

transport process. The VLS grown nanowire array usually possesses better crystallinity than the ones grown by electrochemical deposition in AAO, which offers a platform for making high performance devices. One exceptional application in photovoltaic devices will be discussed in section 4.2.1. However, to avoid deformation of AAO templates, the growth temperature is restricted to $\sim 600^\circ\text{C}$, which sets limitation for materials that can be fabricated by this method.

2.2.2. Vapor-solid mechanism

The vapor-solid (VS) growth is another important method, in which solid nanowires are formed by direct condensation from vapor sources. In the above mentioned VLS growth, unavoidable doping from foreign catalysts may affect the electrical and optical properties. For example, Au is found to induce a deep impurity level in CdTe nanowires^[95]. In contrast, VS growth does not involve metal catalysts, and thus it is preferable when metal impurities in materials cause a major concern. CdS nanowires^[82,96–98], CdSe nanowires^[99] and CdTe nanowires^[90,100,101] have been successfully synthesized by VS growth. However, the fundamental mechanism is not yet well established. Some experimental work has suggested that screw dislocations are responsible for the uni-directional growth^[102,103]. And some have proposed that after the initial nucleation of vapor source, the vapor molecules preferably deposit on the tip due to the Gibbs-Thomson effect, driving the axial growth^[82]. Another possible mechanism suggested is that nanowire growth becomes energetically favorable when the precursor vapor pressure is so low that it is insufficient to facilitate film growth^[98].

2.3. Heterojunction structures

Heterojunction refers to the interface between two dissimilar materials. Semiconductor/semiconductor or semiconductor/metal heterojunctions are of particular interest because their band alignments are fundamentally different from a homojunction, thus providing substantial advantage for many solid state devices such as solar cells, photocatalysts, photodetectors, LEDs (light-emitting diodes), lasers, etc. According to geometries, heterojunction nanowires can be classified into two types: core-shell and axial heterojunction. In both types, obtaining high quality sharp and defect-free interfaces is the critical focus of the research effort.

2.3.1. Core-shell nanowires

In the recent years, core-shell structured nanowires have attracted considerable research interests due to its versatile properties for functional device applications. For example, core-shell nanowire has proven to be a promising structure for highly-efficient photovoltaics based on the main advantages of orthogonalization of light coupling and charge separation directions, shorter diffusion path, as well as enhanced light trapping capabilities. In addition, a shell with larger bandgap can serve as a passivation layer to the core nanowire, leading to better optoelectronic properties.

Core-shell nanowires have been fabricated in one-step or two-step processes. A two-step method is straightforward: first synthesizing core nanowires and then coating a shell layer. For example, CdSe/ZnTe core-shell nanowires have been obtained by thermal evaporation of CdSe nanowires and pulsed laser deposition of ZnTe shell^[104]. The as-obtained CdSe/ZnTe nanowire exhibits a sharp interface with nearly lattice match. And the resulted photovoltaic device will be discussed in the application section 4.2.1. CdS/CdSe core-shell nanowires have been prepared by Au-catalyzed pulsed laser deposition of the core CdS nanowires followed by vapor transport coating of CdSe shell^[105]. This CdS/CdSe nanowires show enhanced optical phonons in resonant Raman scattering. In addition, core-shell CdS:Mn/ZnS nanowires, in which ZnS works as a passivation shell, are fabricated by two-step solvothermal processes: the core CdS:Mn nanowires is first grown in a source bath containing Cd^{2+} and S^{2-} , then is transferred to Zn^{2+} and S^{2-} source solution for the shell growth^[106]. Other two-step processes involve first fabricating homogenous nanowires, and then converting the outer shell into different material via chemical reaction. CdS/CdSe core/shell nanowires have been prepared by fabricating pure CdS nanowires and forming the shell CdSe layers by replacing S with Se^[107]. And Se/CdSe core-shell structures have been obtained by transforming the outer layer of Se nanowires into CdSe by reaction with Cd^{2+} in aqueous media^[108].

One-step growth technique of co-feeding precursors of both types has also been undertaken, which involves more complicated growth kinetics. For instance, Hsu and Lu have reported CdS/ZnS core-shell structure by one-step VLS process. They have attributed the core-shell structure formation to the sufficient reactivity difference between CdS and ZnS, resulting in VLS growth CdS first, and then the CdS serves as catalyst for the successive ZnS shell growth^[109].

2.3.2. Axial heterojunction and superlattice nanowires

Compared with core-shell heterojunctions, axial heterojunctions benefit from relatively simpler fabrication and better controllability. In addition, superlattice nanowires with periodic structure of thin layers of two or more materials are potentially valuable for bandgap engineering. Various prototypes axial heterojunction and superlattice nanowires have been obtained by alternating source precursors in a variety of growth techniques. Schafer et al. have made CdSe/CdTe axial heterojunction nanowires by SLV method. Investigated by electrostatic microscopy, the prepared heterojunction exhibits effective charge separation due to the staggered type-II alignment between CdSe and CdTe, and the charge separation is enhanced under light illumination owing to the photogenerated charge carriers, showing great potential for photovoltaic application^[110]. Bellet-Amalric et al. have demonstrated ZnSe nanowires with CdSe segments obtained by alternately switching Zn flux to Cd flux for 20–45 s in Au-catalyzed MBE growth^[84]. Very short CdSe segments (2–4 nm) are obtained with sharp interfaces, offering a strong confinement in the CdSe segments. Pena et al. have shown metal-CdSe-metal structure by electrochemical deposition in AAO templates, which shows pronounced photoconductivity under visible light illumination^[111]. Furthermore, ZnSe/CdSe superlattice nanowires have been successfully synthesized by atomic layer deposition (ALD) via VLS mechanisms^[112]. ALD offers a precise way to deposit heterojunction structures with monolayer thickness control.

3. Physical Properties—Structural, Electrical and Optical

The device applications of cadmium chalcogenide nanowires strictly depend on their intrinsic material properties. In order to engineer them as the building blocks of nanoscale devices, it is of paramount importance to elucidate their fundamental physical properties. This section will present the basic structural, electrical, and optical properties, respectively for CdS, CdSe, CdTe nanowires.

3.1. CdS nanowires

CdS is one of the most important II–VI semiconductors because of its superior electrical and optical properties. It has a direct bandgap of 2.4 eV ($\lambda \sim 517$ nm), resulting in light emission/absorption in the visible range, thus yielding promising application potentials in LEDs, lasers, photodetectors and other photo-electronic devices. In addition, its piezoelectric property renders important applications in nanogenerators.

Crystal structure of CdS nanowires has been systematically investigated by high resolution electron microscopy (HRTEM) and select area electron diffraction (SAED), as shown in Fig. 2. Single crystalline CdS nanowires are usually found to have wurtzite structure with lattice constant $a = 0.41$ nm and $c = 0.67$ nm, and a preferred growth direction along $[0001]$ (c -axis) arising from the minimum interfacial energy^[36,59,98,113–115]. Fig. 2(a) shows the HRTEM image of a typical CdS nanowire grown along c -axis, illustrating excellent single crystalline characteristics. The interplanar distance ~ 0.67 nm corresponds to the (0001) lattice planes of the wurtzite CdS. The SAED in the inset can be indexed to CdS with zone axis $[\bar{1}2\bar{1}0]$, again confirming the growth direction along $[0001]$. However, owing to variations in fabrication techniques and growth conditions, especially the different crystal orientation of substrate for epitaxial growth, wurtzite CdS nanowires with various growth directions have also been reported^[116,117]. Ma et al. have presented indium doped CdS nanowires fabricated via Au-catalyzed VLS method, and have shown the growth direction along $[\bar{1}1\bar{2}0]$, as illustrated in Fig. 2(b)^[116]. Wu et al. have demonstrated chlorine doped CdS nanowires with single crystalline wurtzite structure with $[\bar{2}\bar{1}\bar{1}1]$ growth direction, synthesized by thermal co-evaporation method, as shown in Fig. 2(c) and (d)^[117]. Moreover, CdS nanowire with less stable zinc blend structure has also been reported, which is fabricated by physical vapor deposition method^[92].

It is essential to study the electrical transport properties of CdS nanowires in order to optimize its performances in electronic and optoelectronic devices. Undoped CdS is typically highly insulating, due to the low intrinsic defect concentration, and thus the electron transport studies of undoped CdS nanowires are rarely reported. In order to tailor the material property and to control the device functionality, doping of CdS nanowires have been extensively investigated. Nanowires are normally configured as field effect transistors (FET), which act as platforms for investigating electrical behaviors and measuring charge carrier concentration and field-effect mobility. A typical nanowire FET is shown in the inset of Fig. 3(a), which consists of a source and a drain electrode with Ohmic contacts and a gate electrode, which is usually separated from the conduction channel by a thin layer of gate dielectrics. More details for an FET configuration will be discussed in section 4.1.1. The field-effect mobility μ and carrier concentration n can be extracted from the electrical characteristics. For example, for a cylindrical nanowires, μ can be determined according to the equations^[119]: $\mu = g_m(L^2/CV_{ds})$, where L is the nanowire channel length; the channel transconductance $g_m = dI_{ds}/dV_{ds}$, where I_{ds} , V_{ds} , are drain-source current and

voltage, respectively. And the gate capacitance $= (2\pi\epsilon_0\epsilon_r L)/[\ln(2h + r)/r]$, where r is the radius of the nanowire, h is the thickness of the gate oxide and ϵ_r is the effective dielectric constant. The carrier concentration n can be estimated from $n = \left(\frac{V_{g,th}}{e}\right)[C/(L\pi r^2)]$, where $V_{g,th}$ is the threshold gate voltage.

n -type gallium doped CdS nanowires have been fabricated by co-evaporation method. And electron concentration and mobility are reported to be around $n = 2.6 \times 10^{17} \text{ cm}^{-3}$ and $\mu = 241.6 \text{ cm}^2/\text{Vs}$, respectively^[120]. Also, n -type indium doped CdS nanowires synthesized via in situ doping CVD method are found to have electron concentration ranging from $1.0 \times 10^{16} - 3.0 \times 10^{17} \text{ cm}^{-3}$, and mobility of $100 - 350 \text{ cm}^2/\text{Vs}$ ^[121]. Recently, electrical transport properties of the amphoteric doping of Sn in CdS nanowire have been investigated^[119]. The $I_{ds} - V_{ds}$ characteristics of a single Sn doped CdS nanowire under different V_g are plotted in Fig. 3(a). I_{ds} increases consistently with increasing V_g , indicating an overall n -type nature. The electron concentration is low, around $8.1 \times 10^{16} \text{ cm}^{-3}$, and the mobility falls in the range of $0.12 - 2.64 \text{ cm}^2/\text{Vs}$. The small mobility validates the self-compensation effect in the nanowire, originating from the amphoteric nature of Sn in CdS. This doping process gives rise to an increased number of ionized donors and acceptors, which contribute to enhanced scattering, leading to the low mobility. Fig. 3(b) shows the temperature-dependent conductivity, σ of Sn doped CdS nanowires, exhibiting three distinct regions. In regions I and III, the conductivity shows Arrhenius-type behavior with $\sigma \propto \exp(-E_a/kT)$, where E_a is the activation energy required to promote an electron into the

conduction band. Note that in compensated semiconductors, the activation energy is simply the ionization energy of the impurity levels. In region III, the activation energy gives $E_D \sim 20 \text{ meV}$, attributing to the shallow donor levels due to the intrinsic and extrinsic dopants. In region I for temperature above 230 K , the activation energy is $\sim 110 \text{ meV}$, contributing to deeper donor level originated from sulfur vacancy. On the other hand, in region II—the intermediate temperature regime, conductivity increases with decreasing temperature at the high temperature side, while inverse behavior is observed at the low temperature side. This exemplifies a characteristic extrinsic saturation region, where all shallow donors are ionized and the carrier concentration in the band is independent of temperature, and the temperature dependence of the conductivity in this region is governed by that of the

mobility. The impurity states and compensation effect are further confirmed in the photoluminescence measurement as discussed next.

Optical properties of CdS nanowires has been intensively studied to reveal the band structure and its unique characteristics. Room temperature photoluminescence of CdS nanowires usually exhibit a single emission peak at ~ 507 nm, originating from near band edge emission (NBE)^[122], and no deep defect band related peaks are observed. Low temperature photoluminescence (PL) ($T < 20$ K) has further revealed exciton peaks localized to morphology defects, however, the microscopic nature is not extensively explored^[123].

Zhang et al. has elucidated a detailed electronic band structure of Sn doped CdS nanowires by cross examining their optical property and the transport characteristics discussed earlier^[119]. Low-temperature ensemble PL ($\lambda_{\text{ex}} = 325$ nm) measurement is plotted in Fig. 4(a), and the corresponding band diagram is depicted in Fig. 4(b). A double structure containing donor acceptor pair (DAP, 2.399 eV) and free-to-bound ((e, A), 2.417 eV) transitions, are most likely to be caused by the incorporation of Sn since these peaks are absent in undoped CdS nanowires. The DAP and (e, A) peaks are strongly accompanied by phonon replica with a distinct spacing corresponding to the energy of a longitudinal optical phonon with $E_{\text{LO}} = 38$ meV. The occurrence of several phonon replica, where the first LO replica is more intense than the zero phonon lines, demonstrates strong electron-phonon coupling. The energetic difference between the DAP and (e, A) peak gives the electron donor binding energy $E_{\text{D}} = 18$ meV. Furthermore, the PL spectra exhibit a richly featured NBE region. The high energetic shoulder of the predominant peak is the free exciton emission FX_{A} at ~ 2.552 eV accompanied by an exciton bound to ionized donors (DX, labeled I_3) at 2.548 eV. The dominant peak near 2.545 eV is a superposition of the exciton bound to neutral donors (D^0X , labeled I_2) and the surface exciton SX. The peaks at the lower energy side are composed by two lines I_1 (2.535 eV) and $\text{I}_{1\text{a}}$ (2.533 eV) corresponding to neutral acceptor-bound excitons and its acoustic phonon replica. The following five peaks correspond to longitudinal optical phonon replicas of $\text{I}_{2\text{B}}$ (B-Exciton bound to neutral donor), FX_{A} , I_2 , and I_1 with respective energy difference of 38 meV.

Also, single CdS nanowire shows significant polarization anisotropic optical properties, that is, the optical emission peaks in Raman or photoluminescence spectra depend highly on the polarization orientation of the incident light^[33,114,124,125]. It is shown that the polarization anisotropy of LO features and their overtones are affected predominantly by the anisotropic shape effect, i.e. the high aspect ratio

and the large dielectric contrast between the nanowires and the surrounding, which can be explain by classical dielectric contrast model^[125]. In essence, Raman scattering is proportional to the square of local internal field intensity. When the polarization of incident light (E_e) is along the nanowires axis, the internal electrical field is $\sim E_e$, however, when the incident field perpendicular to nanowires axis, the internal electric field is reduced to $(2\varepsilon_0/(\varepsilon+\varepsilon_0)E_e)$. Therefore, the angular dependent Raman scattering follows $I(\theta) \propto \cos^2(\theta)E_e^2$. However, the E_2^H band shows a distinctive angular dependence with $I(\theta) \propto \sin^4(\theta)E_e^2$, resulting from the interplay of crystal symmetry, electronic resonance and shape anisotropic effect^[125]. As reported on CdS nanowires: PL peaks is strongest when polarization of incident light is along the nanowires axis and weakest when perpendicular. And the polarization value $r = (I_p - I_\perp) / (I_p + I_\perp)$ is determined to be ~ 0.78 ^[114].

3.2. CdSe nanowires

CdSe has a direct bandgap of 1.74 eV at room temperature. Due to its excellent optical and electrical properties, CdSe nanowires hold promising potential in developing nanoelectronics and optoelectronics, such as light-emitting devices, photodetectors and photovoltaics.

CdSe nanowires have typically stable wurtzite structure with lattice constant $a_0 = 0.43$ nm, $c_0 = 0.70$ nm and preferred growth direction along [0001] (c -axis)^[77,81,126]. However, the growth direction depends on the orientation of substrates and doping incorporations. As shown in Fig. 5, CdSe nanowires are found to grow exclusively along the $[10\bar{1}0]$ direction on GaAs (100) substrate^[81]. In contrast, CdSe nanowires grown on GaAs (111)B plane has growth direction along the [0001] direction instead, as shown in Fig. 5(b). The difference may be attributed to the lattice mismatch between the substrate crystal plane and the nanowires. Furthermore, ultra narrow CdSe nanowires fabricated by SLS method commonly manifested a mixture of zinc blend and wurtzite structure, resulting from the small energy difference between these two phases in nanowires with very small diameters^[127]. And these narrow CdSe nanowires prefer to grow along either zinc blend [111] or wurtzite [0001] direction^[127].

Fundamental electrical and optical properties have been explored to evaluate CdSe nanowires as basic components for nanoelectronic and optoelectronic applications^[128]. Unintentionally doped CdSe nanowires exhibit very low conductivity on the order of 10^{-4} S/cm due to low intrinsic defect density. A

typical I_{DS} – V_{DS} characteristic curve of undoped CdSe nanowire is shown in Fig. 6(a). At applied voltage of 2 V, the detected current is as low as ~20 pA. In order to improve the electrical and photoelectric properties, indium doping on CdSe nanowires has been successfully obtained via in-situ doping or post doping method in CVD chamber. Indium doping effectively enhances the conductivity of CdSe nanowires by four orders of magnitude, as shown in Fig. 6(b). And electron concentration as high as $\sim 10^{19} \text{ cm}^{-3}$ and mobility $\sim 6.7 \text{ cm}^2/\text{Vs}$ are estimated^[128].

Optical properties of CdSe nanowires have also been examined by a few groups^[81,99,129]. Low temperature photoluminescence spectra of CdSe nanowires are measured at 10 K, as shown in Fig. 7^[99]. CdSe nanowires without morphology defects show only NBE emission at 1.815 eV (Fig. 7(a)), while those with morphology defects also exhibit O' peak at ~1.71 eV and O'' peak at ~1.75 eV, which implies O-band (O' and O'') originates from extended morphology defects (Fig. 7 (b)). O' and O'' peaks arise from DAP recombination related to different defect states, confirmed by power-dependent PL measurement. NBE emissions of Au-catalyzed nanowires (Au-NW) and seeded-grown nanowires (SG-NW, without catalysts) are compared, as depicted in Fig. 7(c). NBE emission of SG-NW exhibits a free exciton peak at 1.825 eV (labeled as S), two bound exciton peaks at 1.821 eV (labeled as U) and 1.818 eV (labeled as V), giving an overall NBE with FWHM ~7 meV. In contrast, NBE emission of Au-NW displays three peaks: W (1.826 eV, free exciton), Y (1.816 eV, bound exciton) and Z (1.809 eV, donor recombination), giving an overall FWHM ~30 meV. A smaller FWHM of NBE suggests better crystal quality for the SG-NW, presumably due to higher temperature in seeded growth (~700 °C) compared with that in Au-catalyzed growth (~450 °C). Additionally, significant polarization anisotropic PL in individual CdSe nanowires due to the shape anisotropy are also revealed^[81]. Fig. 8(a) shows the PL intensity as a function of the excitation polarization angle θ (the angle between the polarization of the incident light and the nanowire axis), which fits well to $\cos^2\theta$ dependence. And the photoluminescence spectra for excitation parallel (blue) and perpendicular (red) to the nanowire axis are shown in Fig. 8(b), suggesting stronger coupling when the polarization is parallel to the nanowire axis.

3.3. CdTe nanowires

Cadmium telluride (CdTe) has attracted enormous research interest because of its exceptional optical and electrical properties for developing optoelectronic devices. In particular, it holds promising potential for low-cost photovoltaics, due to its high absorption coefficient and a direct band gap of 1.49 eV at

room temperature, which is close to the theoretically calculated optimal value for solar cells under AM 1.5 irradiation. Vapor grown CdTe nanowires are single crystalline zinc blend structure with lattice constant $a_0 = 0.648$ nm in most circumstances, as revealed by electron microscopy and electron diffraction studies. Fig. 9(a) shows TEM image of a single CdTe nanowire fabricated by CVD method. The metal catalyst present on the tip suggests a catalytic growth mechanism. The corresponding HRTEM image in Fig. 9(b) shows lattice fringes, and the interplanar distance is measured to be 0.37 nm, corresponding to the [111] direction in zinc blende CdTe. These are confirmed by the SAED pattern displayed in the inset^[95].

In addition, wurzite CdTe nanowires with kinks are also fabricated. The growth directions of the segments are respectively along [111] and [100] direction, as labeled in Fig. 9(c)^[87]. Such configuration may lead to novel device design and applications. Moreover, Luo et al. have prepared both zinc blend and wurzite structure CdTe nanowire arrays by sputtering method^[101]. It is reported that pure zinc blende phase nanowire arrays that stack along the [111] direction are obtained at a high growth temperature with a low deposition rate. Conversely, wurzite phase CdTe nanowire arrays with growth in the [0002] direction are obtained at a lower growth temperature with a higher deposition rate. A Gibbs free energy nucleation model is applied to explain the formation of these different crystal phases under the employed growth conditions. Furthermore, very narrow CdTe nanowires fabricated by SLS mechanism are commonly found to be admixtures of zinc blend and wurzite phases, similar to the CdS nanowires obtained via SLS method^[130].

Electrical and optical properties of CdTe nanowires have been investigated. Undoped CdTe nanowires exhibit high resistivity^[87] with low hole concentration presence due to intrinsic defects such as Cd vacancies (V_{Cd}) and defect complexes involving V_{Cd} . To gain a comprehension on the impurity states in undoped CdTe nanowires, low temperature photoluminescence properties of undoped CdTe nanowires have been investigated and the electronic band structure is determined and shown in Fig. 10^[95]. Apart from the sharp bound exaction peak ~ 1.598 eV (peak 1), other impurity related peaks are also revealed. The peak centered around 1.55 eV is in fact a double peak (peak 2 and 3), consisting of a donor–acceptor pair (DAP) transition channel and a free electron to acceptor (e, A) transition channel related to the same acceptor energy level. The corresponding acceptor state is located at ~ 50 meV above the valence band edge, and is likely to originate from the defect state due to Cd vacancies (V_{Cd}) or defect complexes involving V_{Cd} . The 1.494 eV peak (peak 4) can be identified as the well-known Y

luminescence band, originating from the excitons bound to the structural dislocations. The peak at 1.335 eV (peak 5) can be attributed to a DAP transition involving Au acceptors (with energy level E_{Au} about 263 meV above the valence band) and a shallow donor located a few meV below the conduction band, originating from Te antisite (Te_{Cd}). It is also observed from the PL spectra that the exciton and impurity energy levels are shifted and broadened, signifying residual strain effect in the as-grown CdTe nanowires.

In order to enhance the electronic property and device performance, efficient doping is critical, and the preserving difficulty lies in achieving low resistivity as well as high mobility. So it is important to understand the fundamental mechanisms of impurity doping. Among the work on CdTe nanostructures doped by $\text{Zn}^{[131-133]}$, $\text{Sb}^{[9,134,135]}$, or $\text{Mn}^{[136]}$, Sb has been proven to be an effective *p*-type dopant. Controlled *p*-type doping has been achieved by co-evaporation of Sb and CdTe^[135]. As shown in Fig. 11, the conductivity increases by two orders of magnitude after Sb doping. The hole concentration is measured to be around 10^{16} – 10^{17} cm^{-3} even with high concentration of Sb dopants. This attributes to the self-compensation of acceptor impurities. Huang et al. have recently investigated the nature of Sb dopants in CdTe nanowires, and have found surprisingly that in addition to shallow acceptor state, Sb induces deep level AX centers, resulting in significant compensations, manifested by the low carrier concentration and low mobility even at high Sb incorporation^[137]. From cross examination of temperature-dependent charge transport and photoluminescence measurements, two major acceptor levels induced by Sb doping are determined: a shallow level attributed to substitutional Sb dopants without lattice relaxation and an associated deeper level resulted from large lattice relaxation, so called AX centers. Persistent photoconductivity and hysteresis photoconductance under thermal cycle verify the existence of AX centers^[137].

Low-temperature PL spectrum of Sb doped CdTe nanobelt has also been investigated. Fig. 12 shows the PL spectrum taken at 3.7 K of a CdTe:Sb nanobelt ensemble and the recombination processes are depicted in the schematic band diagram. The luminescence of free exciton transitions (FX) is energetically found at ~ 1.596 eV in agreement with the bandgap energy of 1.609 eV reduced by the exciton binding energy (E_{B}^{FX}) of 13 meV^[138]. Two phonon replica of the FX arise at ~ 1.576 eV and ~ 1.555 eV lowered by the phonon energy (E_{LO}) of 21 meV^[138]. A donor bound exciton (D, X) peak is found at 1.594 eV and is attributed to Te antisites (Te_{Cd})^[139,140], while the 1.590 eV line is related to the recombination of an acceptor bound exciton (A, X)^[138]. Furthermore, at lower energies several

recombination signatures occur. The (e, A) transition is induced by a free electron recombining with an acceptor bound hole, while the donor–acceptor pair (DAP) transition is the emission feature of an electron bound to a donor recombining with a hole bound to an acceptor. The higher energetic (e, A)₁ transition and the corresponding DAP₁ transition are located at 1.548 eV and 1.539 eV. Since both rely on the same acceptor level, the energy difference between these two series corresponds to the binding energy of an electron to a donor, $E_D = 9$ meV. The involved acceptor level exhibits an ionization energy of 60 meV, which coincides very well with the shallow acceptor level determined by the electrical measurements. The 1.500 eV and 1.492 eV emission lines for Sb doped CdTe reported in their work is pioneering. These emissions are attributed to a second pair of (e, A)₂ and DAP₂ transitions due to the deeper acceptor level at 109 meV above the valance band edge, which correspond well to their electrical measurement with a comparable activation energy of 115 meV. Moreover, the most intense luminescence around 1.450 eV, which is accompanied by its phonon replicas, is labelled as DAP₃. The corresponding acceptor levels are related to the A center^[141]. The impurities levels are revealed. In brief, a band diagram of the impurity levels the donor level ($E_D = 9$ meV) is attributed to Te antisites. The deepest acceptor level ($E_{A3} \sim 130$ meV) is related to the A centers, and two acceptor levels at $E_{A1} \sim 60$ meV and $E_{A2} \sim 110$ meV are induced by the Sb doping, while E_{A1} is by substitutional Sb atom on a Te site, and E_{A2} is originated from AX-centers.

4. Applications

Due to their excellent electrical, optical and piezoelectric properties, cadmium chalcogenide nanowires offer a wealth of device applications. This section will feature the broad applications based on these nanowires in nanoelectronic, optoelectronic, and electromechanical devices.

4.1. Electronic components

4.1.1. Field-effect transistors (FET)

Nanowire FET has drawn considerable attention in the research community. It not only provides a convenient way to study the fundamental electrical transport and light coupling properties, but also demonstrates substantial advantages serving as a modular component for hybrid electronic systems, such as logic gates, memory devices, photodetectors and sensors. FETs based on single CdS^[116,120], CdSe^[128,142] and CdTe nanowires^[135] have been studied by many groups. The fabrication process is

standard: photolithography or electron-beam lithography is employed to define source and drain electrodes on single nanowires, followed by configuring either a bottom or top gate electrode. The back gated FET utilizes degenerately doped Si ($n++$ or $p++$) substrate as the gate electrode; and the top gated structure requires more fabrication procedures including gate dielectric layer deposition followed by gate electrode deposition. Both back gated and top gated CdS:Ga nanowire FETs were investigated and their performances were compared^[120]. Fig. 13(a) illustrates the schematic diagram of a back gated CdS:Ga nanowire FET, and Fig. 13(b) exhibits its $I_{DS}-V_{DS}$ and $I_{DS}-V_g$ characteristics, showing small ON/OFF current ratio of 1.3 at large operation voltage $V_G = 40$ V, and low electron mobility of $2.1 \text{ cm}^2/\text{Vs}$. Fig. 13(c) and (e) show an SEM image and schematic diagram of a top gated CdS:Ga nanowire FET, and Fig. 13 (d) and (f) depict its $I_{DS}-V_{DS}$ and $I_{DS}-V_g$ characteristics. The top gated FET has shown significantly improved performances: the ON/OFF current ratio reaches as high as $\sim 10^7$ in the gate voltage range from -2 V to 3 V, and the electron mobility increases to $241.6 \text{ cm}^2/\text{Vs}$.

Most reported nanowire FETs are MISFET (metal-insulator-semiconductor) operated at depletion mode (D-mode), which works at ON state at zero gate voltage. However, enhancement-mode (E-mode) FETs usually have more advantages for high speed and low power devices. E-mode FET based on single CdS:In nanowires is produced by fabricating a top surrounding Au layer as a Schottky gate^[116]. Unlike the MISFET structure, this MESFET (metal-semiconductor) structure has no gate dielectric between the gate metal and the nanowire channel. At zero gate voltage, if the nanowire is small enough, the build-in potential of the Schottky junction is sufficient to deplete and pinch off the channel. As gate voltage increases from 0 V to 0.25 V, I_{DS} increases exponentially due to the shrinking of the depletion region that opens up the channel for carrier conduction. A maximum transconductance of $\sim 14.6 \text{ nA/V}$ and an ON/OFF current ratio of 5×10^3 have been obtained.

4.1.2. Logic circuits

The fundamental FET component with sharp ON/OFF switching states can be assembled into complex logic circuits, such as NOT, NOR, and NAND gates, for developing integrated nanoelectronics. High performance NOT, NOR, and NAND logic circuits based on CdS MESFET component have been constructed, as displayed in Fig. 14^[143]. The NOT logic gate (inverter) is constructed by combining two D-mode MESFET as shown in Fig. 14(a). One of the MESFET is used as a switch and the other is used as a load. When input = 1, the bottom MESFET is turned on and grounded the output, yielding output =

0. When input = 0, the bottom MESFET is turned off, and positive voltage supply leads to output = 1. The transfer characteristics and the corresponding diagram are depicted in Fig. 14(b). Moreover, by combining three identical MESFETs, NOR and NAND logic circuits are demonstrated. The circuit diagram and the corresponding output voltage are plotted in Fig. 14(c) and (d), respectively.

4.1.3. Field-emission components

Field emission, also known as field electron emission, is discharge of electrons induced by an electrostatic field. There is high demand for high quality field emission component in a wide range of applications such as flat panel display, parallel beam electron microscope, X-ray generator, etc. Because of the nanometric dimensions, nanowires are found to be ideal for building field emission cathode, which gives high current density at a lower applied potential as compared to the conventional field emitter counterparts. Since CdS has an electron affinity of 4.2 eV, which is much lower than the generally used nanowire field emitter, e.g. C (5.0 eV) and ZnO (5.3 eV), therefore, CdS nanowires is expected to be a promising field emitter with lower turn-on field and higher current density. Aligned CdS nanowire array with high packing density has shown further improved field emission properties^[23,98]. Lin et al.^[98] have synthesized of CdS nanowire array by MOCVD method, which exhibits a turn-on field ~ 0.68 V/ μm and high field emission current density of 225 mA/cm² at an applied electric field of 20 V/ μm . Datta et al. have fabricated aligned CdS nanowires by solvothermal method, and the turn-on electric field of 0.68 V/ μm is obtained. They have also shown that the field emission current enhances by 50% upon light illumination, which is attributed to photo-generated charge carriers^[23]. By simple thermal evaporation technique, Chavan et al. have synthesized ultra-long CdS nanowires with high aspect ratio ($\sim 8 \times 10^4$), which yields superior field emission properties along with a turn-on field as low as 0.17 V/ μm ^[74].

4.2 .Optical and optoelectronic components

4.2.1. Photovoltaics

CdTe/CdS thin film solar cell has been commercialized for a few decades. The implementation of traditional planar structure solar electricity is always limited by its high cost associated with the inefficient carrier collection. In such planar structure, light absorption and charge carrier collection are in the same direction, thus, in order to ensure both light absorption and charge collection efficiency,

materials with high purity and sufficient thickness are required. This problem can be overcome by introducing a new architecture design, i.e. core-shell nanostructure. It not only orthogonalizes the directions of light absorption and charge separation, but also reduces the active material volume and enhances light trapping capability. Yet because of the enlarged surface area, the major challenge for making highly efficient core-shell structure solar cell is to improve the interface properties in order to minimize interface recombination. Wang et al. have prepared core-shell CdSe/ZnTe nanowire, as discussed in the synthesis section^[104], and demonstrated its photovoltaic property. Without surface optimization, an energy-conversion efficiency of ~1.7% was achieved^[104]. Vertical nanowire array provides a convenient template for constructing large scale core-shell structure solar cells. Fan et al. have demonstrated a 3D nanopillar flexible solar cell with high performances^[93]. High quality CdS nanowire array is first deposited into AAO template via VLS growth, then p-CdTe is coated to form CdS/CdTe heterojunction solar cell. And the whole structure can be readily transferred to another substrate such as a flexible PMDS, as shown in Fig. 15(a), and the J - V characteristics are shown in Fig. 15(b). An efficiency of ~6% is achieved with an open circuit voltage $V_{oc} = 0.62$ V, a short circuit current density $J_{sc} = 21$ mA/cm² and a fill factor $FF = 0.43$ under AM 1.5 G illumination. Furthermore, semiconductor heterojunction photovoltaics, organic/inorganic hybrid photovoltaic cells fabricated with CdSe nanowires and PEDOT:PSS have obtained performance with $J_{sc} = 8.08$ mA/cm², $V_{oc} = 0.642$ V and $FF = 0.69$, and an overall energy conversion efficiency $\eta = 3.6\%$ ^[126].

4.2.2. Photodectors

Nanowires are prominent material for constructing photodectors with high sensitivity owing to their large surface-to-volume ratio. And due to its small scale, it has promising potential for integrating into CMOS technology. Despite various device architectures that have been proposed for nanowire photodectors, the basic principle for photodetection lies in the photoconductivity of semiconductor materials. Collecting photogenerated carriers provides an effective approach for photodetection. CdTe, CdSe and CdS nanowires with direct bandgaps of 1.5, 1.7 and 2.4 eV, respectively, are widely employed for light detection ranging from infrared to ultraviolet range. The simplest device architecture is configured as nanowire FET^[10,87,117,128,144]. A CdTe nanowire photodetector with high responsivity ($R \sim 780$ A/W) and high gain (2.4×10^3) at 400 nm illumination is reported, however, slow decay ($\tau \sim 17$ s) is observed due to the presence of deep level traps in the bandgap^[8]. Optimizing the response speed while

keeping high gain and responsivity requires high material quality with low defect density. Following this principle, Ye et al. have demonstrated CdTe nanowire photodetector with fast response time (< 0.1 s) and a reasonable responsivity ($R \sim 19.2$ A/W), attributing to the low defect density in high quality undoped CdTe nanowires fabricated via VLS method^[87]. Also, single crystalline CdS:Cl nanowires are fabricated which shows fast response speed (< 1 s) as shown in Fig. 16(a), demonstrating superior photoelectric properties with responsivity achieving $R \sim 3 \times 10^3$ A/W and gain as high as 7.8×10^3 ^[117]. Moreover, due to the anisotropy of light absorption, the photocurrent varies with the polarization angle of the incident light θ as $I_{PC} = I_{PC}^0 \cos^2 \theta$ ^[40,117]. Fig. 16(b) depicts a schematic diagram of the device setup for measuring the polarization dependence of a CdS:Cl nanowire photodetector. And the photoconductivity as a function of polarization angle is shown in Fig. 16(c), illustrating strong polarization sensitivity in CdS:Cl nanowire photodetection^[117]. Various strategies are proposed to further improve the performances of nanowire photodetectors. For example, Luo et al. have exemplified that Au nanoparticle decoration can effectively enhance the photodetection sensitivity of CdTe nanowire photodetectors, which is attributed to the local surface plasmon resonance which efficiently traps incident light^[144].

4.2.3. Waveguides and lasers

Due to high refractive index contrast between nanowire and its surroundings, nanowire can function as effective optical cavities which show prominent optical waveguide properties. Waveguide behavior of CdS nanowires has been intensively investigated^[113,145–147]. Barrelet et al. have studied the optical waveguiding in CdS nanowires by scanning optical microscopy (SOM). Under laser excitation, part of the photoluminescent light is guided by the nanowire channels. SOM image recorded on straight nanowires suggests that there is no loss in the channels within the measurement limit^[145]. Superior optical waveguide behavior is also observed in ultra-long CdS nanowires fabricated by Sn-catalyst VLS method^[146]. Unavoidable losses in nanowire waveguide lies in the self-absorption induced by impurity states in nanowires, thus high quality with minimum defect density is usually required for achieving waveguide with low loss. However, defect states could also promote the waveguide properties in some cases. For example, Xu et al. have illustrated diluted Sn-doped CdS nanowires with improved guiding efficiency, as shown in Fig. 17^[147]. Photoluminescence spectra of CdS:Sn nanowires show predominately a broad trap-state emission peak spanning from 540–720 nm and a band edge emission

peak at ~520 nm. The relative intensity of trap state emission and band edge emission peak increases with increasing Sn incorporation, suggesting the trap-state is related to Sn dopants. It is illustrated that the Sn doped CdS nanowire waveguides have lower loss than undoped ones. As shown in Fig. 17, bright luminescence spots are observed at the end of the CdS:Sn nanowires and there is no obvious light leakage along the nanowires, suggesting excellent guiding property. Also, along the propagating distance, the band edge emission decreases much faster than the trap-state emission, proving the effectiveness for guiding the trap-state emission. The loss mechanism is evident in the absorption spectrum, where strong absorption occurs at energies above the bandgap (2.42 eV), with very weak absorption at low energies (550–800 nm). Therefore, the band edge emission subjects to self-absorption. In contrast, the trap state emission is immune to self-absorption, thus manifests high waveguide efficiency.

Based on the effective waveguiding property, CdS nanowire serves as Fabry–Perot nanolasers with its strong optical mode confinement. Optical pumped and electrical pumped lasers based on CdS nanowires have been extensively studied^[2,11,12,51,97,113,148]. Fig. 18 demonstrates the first electrically pump nanowire^[148]. The devices are constructed by assembling CdS nanowires on heavily doped *p*-Si on insulator substrates. In this structure (Fig. 18(a)), electrons and holes can be injected into the CdS nanowire along the whole length from either the top metal layer or the bottom *p*-Si layer. Optical and electrical measurements made on single-crystal CdS nanowires show that these structures function as Fabry–Perot optical cavities with mode spacing inversely proportional to the nanowire length. On the other side, simulated emission from CdS nanowires by optical pumping is observed, the result indicates that both electron-hole plasma (EHP) and Fabry–Perot processes are involved^[97,113]. Geburt et al. have demonstrated CdS laser with low threshold of 10 kW/cm² at room temperature^[11]. By studying the geometrical dependence of lasing action, it is found that the minimum requirement for nanowire diameter can be generalized to be $\sim \lambda/n$, where λ is the wavelength and n is the index of refraction. In doped CdS nanowires, continuous wave laser action in single CdS:Sn nanowires is demonstrated, as shown in Fig. 19^[12]. The s-shape observed in the log-log plot of pump intensity versus optical power (Fig. 19 (b)) clearly confirms the lasing action of CdS nanowires with extreme low lasing threshold of ~2 kW/cm². The optical gain mechanism is most likely caused by an electron-hole plasma upon high pumping powers.

4.3. Photocatalysts

Photocatalytic water splitting to produce hydrogen using solar energy has been considered as one of the most important technology that could provide a solution to the global energy crisis. Over the past decades, several metal oxides including TiO_2 , $\text{La}_2\text{Ti}_2\text{O}_7$ and $\text{K}_2\text{La}_2\text{Ti}_3\text{O}_{10}$ have shown high photocatalytic activities for splitting water under UV light illumination. However, UV light only accounts for 4% of the solar spectrum. Therefore, it is highly desirable to develop photocatalysts that can be activated under visible light region. CdS is a remarkable material for visible light driven photocatalytic hydrogen generation, because of its direct bandgap of 2.4 eV and ideal band position which can facilitate both oxidation and reduction of water. Moreover, large surface-to-volume ratio of CdS nanowire can also promote the photo-efficiency and stability. Jang et al. have reported the photocatalytic performance of CdS nanowires fabricated by solvothermal method^[149]. The photocatalytic reaction is carried out in an aqueous electrolyte solution containing 0.1 mol/L Na_2S and 0.02 M Na_2SO_3 as sacrificial reagent for Pt-loaded (~1.0 wt%) CdS catalysts. It is found that CdS nanowires with higher crystallinity show a higher rate of photocatalytic hydrogen production. Heterojunction structure is expected to be effective for improving photocatalytic activity through better charge separation. For instance, Jang et al. have fabricated CdS nanowires decorated with TiO_2 nanoparticles for photocatalytic hydrogen, which exhibit higher rate of hydrogen production than that of undecorated CdS nanowires^[150]. To optimize photocatalytic activity, the molar concentration of TiO_2 nanoparticles has to be controlled. More TiO_2 nanoparticle coverage benefits charge separation at the CdS/ TiO_2 interface; whereas the decreased CdS/solution interface inhibits hole transport through the solution. It is observed that the optimal value for TiO_2 nanoparticle coverage is ~0.2 at% by balancing the CdS/ TiO_2 interface for charge separation and the CdS/solution interface for hole scavenging. Moreover, Jin et al. have demonstrated CdS nanowires decorated with Pt nanoparticles (0.3 wt%) yielding H_2 -production rate as high as 1.49 mmol/h, with a remarkable quantum efficiency of 61.7%, according to the effective charge separation in the CdS/Pt interface^[151]. Other heterostructure are also proposed, such as the CdS nanowires and reduced graphene oxide nanocomposites, which exhibit higher photocatalytic activity than pure CdS nanowires.

4.4. Piezoelectric nanogenerator

Nanogenerators are nanostructured energy harvesters that convert mechanical energy of vibrational and mechanical sources into electrical power, with the potential for replacing external circuits or batteries in integrated nanoelectronics^[152,153]. Nanowire piezoelectric nanogenerator has been demonstrated to effectively convert mechanical energy into electricity. The piezoelectric property of CdS renders it a promising candidate for such energy conversion devices^[154]. Piezoelectric nanogenerators based on CdS nanowires arrays have been fabricated using hydrothermal and physical vapor deposition (PVD) methods. The piezoelectric measurement has been carried out in contact mode of AFM using a Pt coated Si tip. When scanning the tip across the sample, the bending of the nanowires induces a voltage, which is detected by an external load. Importantly such response depends on the crystal structure. Since wurtzite structure is piezoactive but zinc blend structure is not, it is important to utilize CdS nanowire with pure wurtzite phase for optimizing the performance of CdS nanowire piezoelectric nanogenerators.

5. Conclusion and Outlook

This article provides a comprehensive review of the current research status on cadmium chalcogenide nanowires, ranging from synthesis, properties, and device applications. Due to the tremendous effort devoted, one article could hardly cover the extent of all fascinating work that has been reported. Synthesis techniques have been well matured and high quality cadmium chalcogenide nanowires have been successfully fabricated. Based on the excellent physical properties of the cadmium chalcogenide nanowires, they have provided a playground for both fundamental science studies as well as technological advancement. Significant endeavor has been undertaking in device applications utilizing cadmium chalcogenide nanowires, including but not limited to field-effect transistors, logic circuits, field-emission components, photovoltaics, photodetectors, waveguides and lasers. These unprecedented achievements inspire more research interest to address the challenges that remains. (1) Reliable control of the morphology, doping, heterojunction fabrication for functional device application is still in an early stage of development. (2) Comprehensive in-depth understanding and quantitative analysis of the cadmium chalcogenide nanowires still await elucidation. For example, the quantum confinement size effects in the cadmium chalcogenide nanowires are not yet well understood, mainly due to the difficulty

of making high quality narrow nanowires comparable to the Bohr radius. Also, the electron transport mechanism in the material system is seldom reported, though it is vital for electronic devices applications. Moreover, the interplay/coupling of electrons, spin, photons and phonons demand further investigation. (3) Large-scale device integration remains a crucial challenge. The realization of logic circuits based on cadmium chalcogenide nanowires shows a great leap forward. However, cascading them into multilevel combinational logic circuits with high gain and efficient interconnection for hierarchical assembly still requires much investigation. Last but not the least, all the innovative discoveries and inventions will continuously stimulate research curiosity and creativity, to address the technical challenges, to unravel the scientific puzzles, and to develop and design nanoscale devices beyond those one can imagine.

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References

- [1] S. Iijima, *Nature* 354 (1991) 56–58.
- [2] R. Robert, P. Daniel, K. Arian, B. Robert, G. Sebastian, P. Ulf, R. Carsten, *J. Phys. D-Appl. Phys.* 47 (2014) 394012.
- [3] Z. Fan, D. Wang, P.C. Chang, W.Y. Tseng, J.G. Lu, *Appl. Phys. Lett.* 85 (2004) 5923–5925.
- [4] C. Blömers, J.G. Lu, L. Huang, C. Witte, D. Grützmacher, H. Lüth, T. Schäpers, *Nano Lett.* 12 (2012) 2768–2772.
- [5] Z. Fan, J.G. Lu, *Appl. Phys. Lett.* 86 (2005) 123510.
- [6] P.C. Chang, Z. Fan, C.J. Chien, D. Stichtenoth, C. Ronning, J.G. Lu, *Appl. Phys. Lett.* 89 (2006) 133113.
- [7] L. Huang, D. Li, P. Chang, S. Chu, H. Bozler, I.S. Beloborodov, J.G. Lu, *Phys. Rev. B* 83 (2011) 245310.
- [8] X. Xie, S.Y. Kwok, Z. Lu, Y. Liu, Y. Cao, L. Luo, J.A. Zapien, I. Bello, C.S. Lee, S.T. Lee, W. Zhang, *Nanoscale* 4 (2012) 2914–2919.

- [9] C. Xie, L.B. Luo, L.H. Zeng, L. Zhu, J.J. Chen, B. Nie, J.G. Hu, Q. Li, C.Y. Wu, L. Wang, J.S. Jie, *Crystengcomm* 14 (2012) 7222–7228.
- [10] G. Yang, B.-J. Kim, D. Kim, J. Kim, *Opt. Express* 22 (2014) 18843–18848.
- [11] S. Geburt, A. Thielmann, R. Roder, C. Borschel, A. McDonnell, M. Kozlik, J. Kuhnel, K.A. Sunter, F. Capasso, C. Ronning, *Nanotechnology* 23 (2012) 365204.
- [12] R. Röder, M. Wille, S. Geburt, J. Rensberg, M. Zhang, J.G. Lu, F. Capasso, R. Buschlinger, U. Peschel, C. Ronning, *Nano Lett.* 13 (2013) 3602–3606.
- [13] Z.Y. Fan, H. Razavi, J.W. Do, A. Moriwaki, O. Ergen, Y.L. Chueh, P.W. Leu, J.C. Ho, T. Takahashi, L.A. Reichertz, S. Neale, K. Yu, M. Wu, J.W. Ager, A. Javey, *Nat. Mater.* 8 (2009) 648–653.
- [14] D. Routkevitch, T. Bigioni, M. Moskovits, J.M. Xu, *J. Phys. Chem.* 100 (1996) 14037–14047.
- [15] D.S. Xu, Y.J. Xu, D.P. Chen, G.L. Guo, L.L. Gui, Y.Q. Tang, *Adv. Mater.* 12 (2000) 520–522.
- [16] S.P. Mondal, K. Das, A. Dhar, S.K. Ray, *Nanotechnology* 18 (2007) 095606.
- [17] Y. Zhao, X.C. Yang, W.H. Huang, X. Zou, Z.G. Lu, *J. Mater. Sci.* 45 (2010) 1803–1808.
- [18] Y. Yang, H.L. Chen, Y.F. Mei, J.B. Chen, X.L. Wu, X.M. Bao, *Solid State Commun.* 123 (2002) 279–282.
- [19] J.D. Klein, R.D. Herrick, D. Palmer, M.J. Sailor, C.J. Brumlik, C.R. Martin, *Chem. Mater.* 5 (1993) 902–904.
- [20] H. Zhang, X.Y. Ma, J. Xu, J.J. Niu, J. Sha, D.R. Yang, *J. Cryst. Growth* 246 (2002) 108–112.
- [21] H.Q. Cao, Y. Xu, J.M. Hong, H.B. Liu, G. Yin, B.L. Li, C.Y. Tie, Z. Xu, *Adv. Mater.* 13 (2001) 1393–1394.
- [22] H.Q. Shi, X.D. Zhou, X. Fu, D.B. Wang, Z.S. Hu, *Mater. Lett.* 60 (2006) 1793–1795.
- [23] A. Datta, P.G. Chavan, F.J. Sheini, M.A. More, D.S. Joag, A. Patra, *Cryst. Growth Des.* 9 (2009) 4157–4162.
- [24] L.B. Fan, T.H. Feng, P. Wang, Z.B. Feng, C.L. Zhang, *Aust. J. Chem.* 62 (2009) 448–454.
- [25] S.C. Yan, L.T. Sun, P. Qu, N.P. Huang, Y.C. Song, Z.D. Xiao, *J. Solid. State Chem.* 182 (2009) 2941–2945.
- [26] M.A. Mandi, J.J. Hassan, S.S. Ng, Z. Hassan, *J. Cryst. Growth* 359 (2012) 43–48.
- [27] S.C. Yan, Z.D. Xiao, Y. Shi, D. Hu, X. Xu, J.S. Wu, M.M. Zhou, *Nanosci. Nanotech. Lett.* 5 (2013) 213–221.
- [28] J.G. Zhao, Z.H. Hua, Y. Yao, *Superlattice. Microst.* 61 (2013) 146–151.

- [29] N.M.A. Hadia, S. Garcia-Granda, J.R. Garcia, J. Nanosci. Nanotechnol. 14 (2014) 5449–5454.
- [30] I. Kazeminezhad, N. Hekmat, A. Kiasat, Fiber. Polym. 15 (2014) 672–679.
- [31] M.A. Mahdi, J.J. Hassan, S.J. Kasim, S.S. Ng, Z. Hassan, Bull. Mater. Sci. 37 (2014) 337–345.
- [32] L.P. Jiang, S. Xu, J.J. Miao, H. Wang, J.J. Zhu, J. Nanosci. Nanotechnol. 6 (2006) 2584–2587.
- [33] H.M. Fan, Z.H. Ni, Y.P. Feng, X.F. Fan, J.L. Kuo, Z.X. Shen, B.S. Zou, Appl. Phys. Lett. 91 (2007) 171911.
- [34] G. Murali, D.A. Reddy, G. Giribabu, R.P. Vijayalakshmi, R. Venugopal, J. Alloy. Compd. 581 (2013) 849–855.
- [35] B.S. Simpkins, T. Brintlinger, R.M. Stroud, S. Sherrill, S.B. Lee, P.E. Pehrsson, J. Phys. Chem. C 117 (2013) 11843–11849.
- [36] Z.D. Hu, Y. Yu, H.N. Hu, Mater. Lett. 64 (2010) 863–865.
- [37] Y. Zeng, Y.K. Su, X.M. Wu, J.N. Tang, Rare Metal Mater. Eng. 43 (2014) 413–417 (in Chinese).
- [38] D. Xu, Y. Guo, D. Yu, G. Guo, Y. Tang, D.P. Yu, J. Mater. Res. 17 (2002) 1711–1714.
- [39] A.W. Zhao, G.W. Meng, L.D. Zhang, T. Gao, S.H. Sun, Y.T. Pang, Appl. Phys. A-Mater. Sci. Proc. 76 (2003) 537–539.
- [40] Z. Hackney, L. Mair, K. Skinner, S. Washburn, Mater. Lett. 64 (2010) 2016–2018.
- [41] Y. Li, J.H. Wan, Z.N. Gu, Acta Phys. Chim. Sin. 15 (1999) 1–4.
- [42] Y.J. Xiong, Y. Xie, J. Yang, R. Zhang, C.Z. Wu, G. Du, J. Mater. Chem. 12 (2002) 3712–3716.
- [43] M. Ghenescu, L. Ion, I. Enculescu, C. Tazlaoanu, V.A. Antohe, M. Sima, M. Enculescu, E. Matei, R. Neumann, O. Ghenescu, V. Covlea, S. Antohe, Physica E 40 (2008) 2485–2488.
- [44] E. Matei, N. Preda, M. Enculescu, M. Sima, M. Sima, I. Enculescu, Optoelectron. Adv. Mater. 3 (2009) 1018–1022.
- [45] S.N. Sarangi, S. Rath, K. Goswami, S. Nozaki, S.N. Sahu, Physica E 42 (2010) 1670–1674.
- [46] J. Mallet, I. Kante, P. Fricoteaux, M. Molinari, M. Troyon, J. Solid. State Electron. 16 (2012) 1041–1047.
- [47] I. Enculescu, M. Sima, M. Enculescu, M. Enache, L. Ion, S. Antohe, R. Neumann, Phys. Status Solidi B 244 (2007) 1607–1611.
- [48] F. Wang, A. Dong, J. Sun, R. Tang, H. Yu, W.E. Buhro, Inorg. Chem. 45 (2006) 7511–7521.
- [49] J. Puthussery, A.D. Lan, T.H. Kosel, M. Kuno, ACS Nano 2 (2008) 357–367.
- [50] V. Protasenko, D. Bacinello, M. Kuno, J. Phys. Chem. B 110 (2006) 25322–25331.

- [51] Y.J. Doh, K.N. Maher, L. Ouyang, C.L. Yu, H. Park, J. Park, *Nano Lett.* 8 (2008) 4552–4556.
- [52] Z. Li, A. Kornowski, A. Myalitsin, A. Mews, *Small* 4 (2008) 1698–1702.
- [53] O. Kurtulus, Z. Li, A. Mews, U. Pietsch, *Phys. Status Solidi A* 206 (2009) 1752–1756.
- [54] Z. Li, O. Kurtulus, N. Fu, Z. Wang, A. Kornowski, U. Pietsch, A. Mews, *Adv. Funct. Mater.* 19 (2009) 3650–3661.
- [55] H.S. Jang, J.S. Lee, *Bull. Korean Chem. Soc.* 34 (2013) 590–594.
- [56] N. Reim, A. Littig, D. Behn, A. Mews, *J. Am. Chem. Soc.* 135 (2013) 18520–18527.
- [57] J. Puthussery, T.H. Kosel, M. Kuno, *Small* 5 (2009) 1112–1116.
- [58] Y.M. Sung, W.C. Kwak, T.G. Kim, *Crystengcomm* 14 (2012) 389–392.
- [59] Q.Q. Wang, G. Xu, G.R. Han, *J. Solid. State Chem.* 178 (2005) 2680–2685.
- [60] A. Phuruangrat, T. Thongtem, P. Pongphab, S. Thongtem, *Solid State Sci.* 14 (2012) 1023–1029.
- [61] D. Xu, Z.P. Liu, J.B. Liang, Y.T. Qian, *J. Phys. Chem. B* 109 (2005) 14344–14349.
- [62] Q. Yang, K.B. Tang, C.R. Wang, Y.T. Qian, S.Y. Zhang, *J. Phys. Chem. B* 106 (2002) 9227–9230.
- [63] H.D. Gai, Y.S. Wu, L.L. Wu, Y.C. Shi, M. Jing, K. Zou, *Acta Phys. Chim. Sin.* 24 (2008) 157–160.
- [64] J.H. Zhan, X.G. Yang, S.D. Li, D.W. Wang, Y. Xie, Y.T. Qian, *J. Cryst. Growth* 220 (2000) 231–234.
- [65] J.H. Zhan, X.G. Yang, D.W. Wang, S.D. Li, Y. Xie, Y.N. Xia, Y.T. Qian, *Adv. Mater.* 12 (2000) 1348–1351.
- [66] F. Gao, Q.Y. Lu, D.Y. Zhao, *Chem. Lett.* (2002) 732–733.
- [67] P. Yan, Y. Xie, Y.T. Qian, X.M. Liu, *Chem. Commun.* (1999) 1293–1294.
- [68] U.Y. Jeong, Y.N. Xia, Y.D. Yin, *Chem. Phys. Lett.* 416 (2005) 246–250.
- [69] W.I. Park, H.S. Kim, S.Y. Jang, J. Park, S.Y. Bae, M. Jung, H. Lee, J. Kim, *J. Mater. Chem.* 18 (2008) 875–880.
- [70] Y.W. Wang, G.W. Meng, L.D. Zhang, C.H. Liang, J. Zhang, *Chem. Mater.* 14 (2002) 1773–1777.
- [71] C.J. Barrelet, Y. Wu, D.C. Bell, C.M. Lieber, *J. Am. Chem. Soc.* 125 (2003) 11498–11499.
- [72] L.F. Dong, J. Jiao, M. Coulter, L. Love, *Chem. Phys. Lett.* 376 (2003) 653–658.
- [73] C. Li, Z.T. Liu, Y. Yang, *Nanotechnology* 17 (2006) 1851–1857.
- [74] P.G. Chavan, R.V. Kashid, S.S. Badhade, I.S. Mulla, M.A. More, D.S. Joag, *Vacuum* 101 (2014) 38–45.
- [75] C. Ma, Z.L. Wang, *Adv. Mater.* 17 (2005) 2635–2639.

- [76] G.Z. Dai, Q.L. Zhang, Z.W. Peng, W.C. Zhou, M.X. Xia, Q. Wan, A.L. Pan, B.S. Zou, J. Phys. D- Appl. Phys. 41 (2008) 135301.
- [77] M. Wang, G.T. Fei, Nanoscale Res. Lett. 4 (2009) 1166–1170.
- [78] B.B. Xiao, Y.B. Xu, Physica E 44 (2011) 696–699.
- [79] Y. Yan, Z.M. Liao, Y.Q. Bie, H.C. Wu, Y.B. Zhou, X.W. Fu, D.P. Yu, Appl. Phys. Lett. 99 (2011) 103103.
- [80] C.X. Shan, Z. Liu, S.K. Hark, Appl. Phys. Lett. 90 (2007) 193123.
- [81] C.X. Shan, Z. Liu, S.K. Hark, Phys. Rev. B 74 (2006) 153402.
- [82] C.H. Ye, G.W. Meng, Y.H. Wang, Z. Jiang, L.D. Zhang, J. Phys. Chem. B 106 (2002) 10338–10341.
- [83] K.Y. Lee, J.R. Lim, H. Rho, Y.J. Choi, K.J. Choi, J.G. Park, Appl. Phys. Lett. 91 (2007) 201901.
- [84] E. Bellet-Amalric, M. Eloune-Jamroz, P. Rueda-Fonseca, S. Bounouar, M. Den Hertog, C. Bougerol, R. Andre, Y. Genuist, J.P. Poizat, K. Kheng, J. Cibert, S. Tatarenko, J. Cryst. Growth 378 (2013) 233–237.
- [85] R.S. Wagner, W.C. Ellis, Appl. Phys. Lett. 4 (1964) 89–90.
- [86] M.S. Song, Y. Kim, J. Phys. Chem. C 118 (2014) 5988–5995.
- [87] Y. Ye, L. Dai, T. Sun, L.P. You, R. Zhu, J.Y. Gao, R.M. Peng, D.P. Yu, G.G. Qin, J. Appl. Phys. 108 (2010) 044301.
- [88] S. Neretina, R.A. Hughes, J.F. Britten, N.V. Sochinskii, J.S. Preston, P. Mascher, Nanotechnology 18 (2007) 275301.
- [89] S. Neretina, R.A. Hughes, G.A. Devenyi, N.V. Sochinskii, J.S. Preston, P. Mascher, Nanotechnology 19 (2008) 185601.
- [90] J. Hou, X. Yang, X. Lv, D. Peng, M. Huang, Q. Wang, Appl. Surf. Sci. 257 (2011) 7684–7688.
- [91] S.K.C. Lee, Y. Yu, O. Perez, S. Puscas, T.H. Kosel, M. Kuno, Chem. Mater. 22 (2009) 77–84.
- [92] X.C. Wu, Y.R. Tao, J. Cryst. Growth 242 (2002) 309–312.
- [93] Z. Fan, H. Razavi, J.W. Do, A. Moriwaki, O. Ergen, Y.L. Chueh, P.W. Leu, J.C. Ho, T. Takahashi, L.A. Reichertz, S. Neale, K. Yu, M. Wu, J.W. Ager, A. Javey, Nat. Mater. 8 (2009) 648–653.
- [94] P. Villars, K. Cenzual, H. Okamoto, F. Hulliger, S. Iwata, Pauling File Multinaries Edition- 2012, http://www.springermaterials.com/docs/VSP/datasheet/lpf-c/00956000/LPFC_956113.html, June 2014.
- [95] L. Huang, S. Lu, P. Chang, K. Banerjee, R. Hellwarth, J. Lu, Nano Res. 7 (2014) 228–235.

- [96] S.M. Zhou, *Phys. Status Solidi A* 203 (2006) R45–R47.
- [97] H. Pan, G.C. Xing, Z.H. Ni, W. Ji, Y.P. Feng, Z. Tang, D.H.C. Chua, J. Lin, Z.X. Shen, *Appl. Phys. Lett.* 91 (2007) 193105.
- [98] Y.F. Lin, Y.J. Hsu, S.Y. Lu, S.C. Kung, *Chem. Commun.* (2006) 2391–2393.
- [99] A. Fasoli, A. Colli, F. Martelli, S. Pisana, P.H. Tan, A.C. Ferrari, *Nano Res.* 4 (2011) 343–359.
- [100] X. Wang, J. Wang, M. Zhou, H. Zhu, H. Wang, X. Cui, X. Xiao, Q. Li, *J. Phys. Chem. C* 113 (2009) 16951–16953.
- [101] B. Luo, Y. Deng, Y. Wang, M. Tan, L. Cao, W. Zhu, *Crystengcomm* 14 (2012) 7922–7928.
- [102] H.Z. Zhang, Y.C. Kong, Y.Z. Wang, X. Du, Z.G. Bai, J.J. Wang, D.P. Yu, Y. Ding, Q.L. Hang, S.Q. Feng, *Solid State Commun.* 109 (1999) 677–682.
- [103] M.J. Bierman, Y.K.A. Lau, A.V. Kvit, A.L. Schmitt, S. Jin, *Science* 320 (2008) 1060–1063.
- [104] K. Wang, S.C. Rai, J. Marmon, J.J. Chen, K. Yao, S. Wozny, B.B. Cao, Y.F. Yan, Y. Zhang, W.L. Zhou, *Nanoscale* 6 (2014) 3679–3685.
- [105] J.H. Kim, H. Kim, H. Rho, Y.J. Choi, J.G. Park, *J. Phys. D-Appl. Phys.* 47 (2014) 135104.
- [106] S. Kar, S. Santra, H. Heinrich, *J. Phys. Chem. C* 112 (2008) 4036–4041.
- [107] Y. Xie, P. Yan, J. Lu, Y.T. Qian, S.Y. Zhang, *Chem. Commun.* (1999) 1969–1970.
- [108] X.C. Jiang, B. Mayers, T. Herricks, Y.N. Xia, *Adv. Mater.* 15 (2003) 1740–1743.
- [109] Y.J. Hsu, S.Y. Lu, *Chem. Commun.* (2004) 2102–2103.
- [110] S. Schafer, A. Reich, Z. Wang, T. Kipp, A. Mews, *Appl. Phys. Lett.* 100 (2012) 022110.
- [111] D.J. Pena, J.K.N. Mbindyo, A.J. Carado, T.E. Mallouk, C.D. Keating, B. Razavi, T.S. Mayer, *J. Phys. Chem. B* 106 (2002) 7458–7462.
- [112] R. Solanki, J. Huo, J.L. Freeouf, B. Miner, *Appl. Phys. Lett.* 81 (2002) 3864–3866.
- [113] A.L. Pan, R.B. Liu, Q. Yang, Y.C. Zhu, G.Z. Yang, B.S. Zou, K.Q. Chen, *J. Phys. Chem. B* 109 (2005) 24268–24272.
- [114] S. Acharya, I. Patla, J. Kost, S. Efrima, Y. Golan, *J. Am. Chem. Soc.* 128 (2006) 9294–9295.
- [115] Y.J. Hsu, S.Y. Lu, Y.F. Lin, *Chem. Mater.* 20 (2008) 2854–2856.
- [116] R.M. Ma, L. Dai, G.G. Qin, *Appl. Phys. Lett.* 90 (2007) 093109.
- [117] C.Y. Wu, J.S. Jie, L. Wang, Y.Q. Yu, Q.A. Peng, X.W. Zhang, J.J. Cai, H.E. Guo, D. Wu, Y. Jiang, *Nanotechnology* 21 (2010) 505203.
- [118] Z.T. Liu, C. Li, Y.P. Fu, Y. Yang, *J. Nanosci. Nanotechnol.* 7 (2007) 3152–3156.

- [119] M.Y. Zhang, M. Wille, R. Roeder, S. Heedt, L.B. Huang, Z. Zhu, S. Geburt, D. Gruetzmacher, T. Schapers, C. Ronning, J.G. Lu, *Nano Lett.* 14 (2014) 518–523.
- [120] J.J. Cai, J.S. Jie, P. Jiang, D. Wu, C. Xie, C.Y. Wu, Z. Wang, Y.Q. Yu, L. Wang, X.W. Zhang, Q. Peng, Y. Jiang, *Phys. Chem. Chem. Phys.* 13 (2011) 14663–14667.
- [121] R.M. Ma, L. Dai, H.B. Huo, W.Q. Yang, G.G. Qin, P.H. Tan, C.H. Huang, J. Zheng, *Appl. Phys. Lett.* 89 (2006) 203120.
- [122] W.C. Zhou, D.S. Tang, B.S. Zou, *Physica E* 47 (2013) 162–166.
- [123] T.B. Hoang, L.V. Titova, H.E. Jackson, L.M. Smith, J.M. Yarrison Rice, J.L. Lensch, L.J. Lauhon, *Appl. Phys. Lett.* 89 (2006) 123123.
- [124] H.M. Fan, Z.H. Ni, Y.P. Feng, X.F. Fan, Z.X. Shen, B.S. Zou, *J. Raman Spectrosc.* 38 (2007) 1112–1116.
- [125] H.M. Fan, X.F. Fan, Z.H. Ni, Z.X. Shen, Y.P. Feng, B.S. Zou, *J. Phys. Chem. C* 112 (2008) 1865–1870.
- [126] Z.F. Feng, Q.B. Zhang, L.L. Lin, H.H. Quo, J.Z. Zhou, Z.H. Lin, *Chem. Mater.* 22 (2010) 2705–2710.
- [127] K.L. Hull, J.W. Grebinski, J. Zhang, T.H. Kosel, M. Kuno, in: *Proceedings of SPIE: Conference on Physical Chemistry of Interfaces and Nanomaterials III*, Denver, Colorado, US, August 03-06, 2004.
- [128] Z.B. He, J.S. Jie, W.J. Zhang, W.F. Zhang, L.B. Luo, X. Fan, G.D. Yuan, I. Bello, S.T. Lee, *Small* 5 (2009) 345–350.
- [129] A. Myalitsin, C. Strelow, Z. Wang, Z. Li, T. Kipp, A. Mews, *ACS Nano* 5 (2011) 7920–7927.
- [130] M. Kuno, O. Ahmad, V. Protasenko, D. Bacinello, T.H. Kosel, *Chem. Mater.* 18 (2006) 5722–5732.
- [131] S.M. Zhou, X.H. Zhang, X.M. Meng, S.K. Wu, S.T. Lee, *Appl. Phys. A: Mater. Sci. Proc.* 81 (2005) 1647–1650.
- [132] S.M. Zhou, *Phys. Low-Dimens. Struct.* 2 (2006) 29–33.
- [133] T. Gandhi, K.S. Raja, M. Misra, *Electrochim. Acta* 51 (2006) 5932–5942.
- [134] C. Xie, B. Nie, L. Zhu, L.H. Zeng, Y.Q. Yu, X.H. Wang, Q.L. Fang, L.B. Luo, Y.C. Wu, *Nanotechnology* 24 (2013) 355203.
- [135] L. Zhu, J. Jie, D. Wu, L. Luo, C. Wu, Z. Zhu, Y. Yu, L. Wang, *J. Nanoeng. Nanomanuf.* 2 (2012) 191–196.

- [136] P. Ramasamy, S.I. Mamum, J. Jang, J. Kim, Crystengcomm 15 (2013) 2061–2066.
- [137] L. Huang, C.C. Lin, M. Riediger, R. Röder, P.L. Tse, C. Ronning, J.G. Lu, Nano Lett. 15 (2015) 974–980.
- [138] M. Soltani, M. Certier, R. Evrard, E. Kartheuser, J. Appl. Phys. 78 (1995) 5626–5632.
- [139] M.A. Berding, Phys. Rev. B 60 (1999) 8943–8950.
- [140] K.A. Dhese, P. Devine, D.E. Ashenford, J.E. Nicholls, C.G. Scott, D. Sands, B. Lunn, J. Appl. Phys. 76 (1994) 5423–5428.
- [141] D.M. Hofmann, P. Omling, H.G. Grimmeiss, B.K. Meyer, K.W. Benz, D. Sinerius, Phys. Rev. B 45 (1992) 6247–6250.
- [142] Z.B. He, W.J. Zhang, Y.B. Tang, H.B. Wang, Y.L. Cao, H.S. Song, I. Bello, C.S. Lee, S.T. Lee, Appl. Phys. Lett. 95 (2009) 253107.
- [143] R.M. Ma, L. Dai, H.B. Huo, W.J. Xu, G.G. Oin, Nano Lett. 7 (2007) 3300–3304.
- [144] L. Luo, X. Huang, M. Wang, C. Xie, C. Wu, J. Hu, L. Wang, J. Huang, Small 10 (2014) 2645–2652.
- [145] C.J. Barrelet, A.B. Greytak, C.M. Lieber, Nano Lett. 4 (2004) 1981–1985.
- [146] G.Z. Dai, Q.A. Wan, C.J. Zhou, M. Yan, Q.L. Zhang, B.S. Zou, Chem. Phys. Lett. 497 (2010) 85–88.
- [147] J.Y. Xu, X.J. Zhuang, P.F. Guo, Q.L. Zhang, L. Ma, X.X. Wang, X.L. Zhu, A.L. Pan, J. Mater. Chem. C 1 (2013) 4391–4396.
- [148] X. Duan, Y. Huang, R. Agarwal, C.M. Lieber, Nature 421 (2003) 241–245.
- [149] J.S. Jang, U.A. Joshi, J.S. Lee, J. Phys. Chem. C 111 (2007) 13280–13287.
- [150] J.S. Jang, H.G. Kim, U.A. Joshi, J.W. Jang, J.S. Lee, Int. J. Hydrogen Energ. 33 (2008) 5975–5980.
- [151] J. Jin, J.G. Yu, G. Liu, P.K. Wong, J. Mater. Chem. A 1 (2013) 10927–10934.
- [152] Z.L. Wang, J. Song, Science 312 (2006) 242–246.
- [153] P.X. Gao, J. Song, J. Liu, Z.L. Wang, Adv. Mater. 19 (2007) 67–72.
- [154] Y.F. Lin, J. Song, Y. Ding, S.Y. Lu, Z.L. Wang, Appl. Phys. Lett. 92 (2008) 022105.

Figure Captions

Fig. 1. (a) A schematic diagram of a typical experiment set up for vapor phase growth. (b) A schematic diagram shows key processes in the VLS growth of nanowires. (c) Phase diagram of CdS–Sn shows the alloying, nucleation and growth processes^[94].

Fig. 2. (a) HRTEM image of single crystalline CdS nanowire grown along [0001] direction. Inset shows the corresponding SAED pattern^[118]. (b) HRTEM image of CdS nanowire grown along $[11\bar{2}0]$ direction and the corresponding SAED (inset)^[116]. (c) TEM and SAED of CdS nanowire grown along $[\bar{2}\bar{1}\bar{1}1]$ direction^[117]. (d) HRTEM image of CdS nanowire grown along $[\bar{2}\bar{1}\bar{1}1]$ ^[117].

Fig. 3. Source–drain current I_{ds} versus voltage V_{ds} of a Sn doped CdS single nanowire field effect transistor show good linearity under 0, 20, and 40 V gate voltages. Upper left inset: SEM image of a single wire FET. Lower right inset: source-drain current I_{ds} versus gate voltage V_g plots under $V_{ds} = 2, 4$, and 8 mV, respectively. (b) Temperature dependence of conductivity depicts the thermal activation of electrons from various donor levels^[119].

Fig. 4. (a) Photoluminescence spectra of CdS:Sn nanowire ensemble, taken at 4 K with an excitation density of 50 mW/cm² at 325 nm. (b) The band structure for CdS:Sn determined from the PL peaks.

Fig. 5. (a) SEM image of CdSe nanowires. (b) HRTEM image of CdSe nanowires grown along $[10\bar{1}0]$ direction. (c) HRTEM image of CdSe nanowires grown along [0001] direction^[81].

Fig. 6. (a) A typical I_{DS} – V_{DS} curve of undoped CdSe nanowires. Inset: SEM image of a typical CdSe nanowire FET. (b) I_{DS} – V_{DS} curves at different V_G of In doped CdSe nanowires^[128].

Fig. 7. Photoluminescence ($T = 10$ K) and corresponding TEM images of CdSe nanowires with (a) and without (b) morphology defects. Scale bars: 200 nm. (c) Low temperature PL spectra of Au-catalyst CdSe nanowires (Au-NW) and seeded-grown CdSe nanowires (SG-NW)).

Fig. 8. PL anisotropy in CdSe nanowires. (a) PL emission intensity of individual CdSe nanowires as a function of excitation polarization angle θ . (b) PL spectra obtained with $E_{ex} \perp d$ and $E_{ex} // d$ ^[81].

Fig. 9. (a) TEM image of a single CdTe nanowire grown along [111]. (b) HRTEM image of a CdTe nanowire and the SAED is shown in the inset^[95]. (c) TEM and HRTEM image of a CdTe nanowire with a kink^[87].

Fig. 10. (a) Photoluminescence spectrum of CdTe nanowires measured at 5.0 K. (b) The band structure for CdTe illustrates the transition processes: donor bound excitons (DX), Y band, free electron to acceptor transitions (e, A), and donor–acceptor pair transitions (DAP) involving either shallow acceptors or deep level Au acceptors ^[95].

Fig. 11. (a) I_{ds} – V_{ds} characteristics of an undoped CdTe nanowire at different gate voltages. Inset: the corresponding I_{ds} – V_g curve. (b) I_{ds} – V_{ds} characteristics of a Sb doped CdTe nanowire at different gate voltages. Inset: the corresponding I_{ds} – V_g curve.

Fig. 12. (a) PL spectrum of the CdTe:Sb nanobelt ensemble, taken at 3.7 K with a excitation density of 90 mW/m² at 633 nm. Several near band edge transitions are labeled: the free exciton FX, donor bound excitons (D,X), acceptor bound excitons (A,X) free electron to acceptor (e,A) transition, donor acceptor pair recombination (DAP) and their phonon replica. (b) Schematic band diagram depicts the impurity energy levels within the bandgap and recombination processes. One donor level and 3 acceptor levels are involved, and the binding energies are determined to be $E_D = 9$ meV, $E_{A1} = 61$ meV, $E_{A2} = 109$ meV and $E_{A3} = 130$ meV.

Fig. 13. (a) A schematic illustration of the back-gated CdS:Ga nanowire FET. (b) I_{DS} – V_{DS} curves at different V_G . Inset: the corresponding I_{DS} - V_G curves measured at $V_{DS} = 1$ V. (c) A SEM image of a top-gated CdS:Ga nanowire MISFET. (d) I_{DS} – V_{DS} curves measured at various V_G . (e) A schematic illustration of the top-gated CdS:Ga nanowire FET. (f) I_{DS} – V_G curve measured at $V_{DS} = 1$ V ^[120].

Fig. 14. (a) A SEM image of a NOT logic circuit on one single CdS nanowire. (b) Transfer characteristics of the NOT gate presented in (a), with circuit diagram drawn in the inset. (c) Output voltage of a CdS nanowire NOR gate for the four input/output states: (0, 0), (0, 1), (1, 0), and (1, 1). (d) Output voltage of a CdS nanowire based NAND gate ^[143].

Fig. 15. (a) A schematic diagram of the CdS/CdTe 3D nanopillar structure solar cell on PDMS substrate. (c) J – V characteristics of a flexible solar cell at different bending radii.

Fig. 16. (a) Photoresponse of CdS:Cl nanowires. (b) Schematic diagram shows the measurement configuration for polarization-sensitive measurements. (c) Polarization sensitivity of CdS:Cl nanowire to the incident light with varied polarization angle ^[117].

Fig. 17. (a) SEM image of a single CdS:Sn nanowire, scale bar = 2 μ m. (b) PL images with focused excitation (488 nm) at different positions (E1–E4) along the channel length. Scale bar = 20 μ m. (c) Local PL spectra and their corresponding integrated intensities (inset) recorded at the excitation

positions (E1–E4). (d) Local PL spectra at the output end facets (O1–O4). Inset shows that the spectrum at the O1 distance (red) can be well resolved into a narrow band edge emission and a broad trap-state emission (green)^[147].

Fig. 18. (a) Schematic diagram illustrates the structure of electrically driven CdS nanowire laser. (b) SEM image (top) and electroluminescence image (bottom) of the device. (c) Emission intensity vs. injection current curve. (d) Electroluminescence spectra obtained from the nanowire end facet with injection currents of 120 mA (red) and 210 mA (green). The black arrows highlight Fabry–Perot cavity modes with an average spacing of 1.83 nm^[148].

Fig. 19. (a) The μ PL end facet spectra of a Sn doped CdS nanowire at different excitation power intensities. (b) Log–log plot of the integrated PL intensity data versus pump power. The inset shows the bright continuous wave emission from the single nanowire visible with unaided eye. Inset: bright cw emission from single nanowire visible with unaided eye.

Figure List

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Fig. 3

Fig. 4

Fig. 5

Fig. 6

Fig. 7

Fig. 8

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Fig. 10

Fig. 11

Fig. 12

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