Topical Review

Self-catalyzed GaAs(P) nanowires and their application for solar cells

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Abstract
Nanowires (NWs) are far superior to traditional thin-film counterparts in making high-efficiency low-cost solar cells (SCs) due to their advanced properties, such as reduced light reflection, strong light trapping, increased defect tolerance, and compatibility with Si substrates. This allows us to greatly reduce the quantity of required material and lessen demands on its quality to achieve sufficient light absorption, leading to substantial cost reduction. Self-catalysed III–V NWs are being widely used for SCs, especially GaAs(P) NWs. However, the special growth mode makes the SC design extremely complex and the existing theories cannot be used directly for guidance. Here, we review the recent breakthroughs and remaining challenges in growing and making self-catalysed NW SCs, with special attention given, but not limited, to GaAs(P) NWs.

Keywords: self-catalysed nanowires, solar cells, GaAsP, nanowires

1. Introduction

1.1. Solar cells and remaining challenges
Nowadays, energy consumption by human beings is increasing rapidly. More than 90% of energy is provided by non-renewable fossil fuels, mainly coal, natural gas and petrol, which will run out soon [1]. Moreover, the energy generation by burning those fossil fuels is accompanied by many side effects. The exhaust contains a large amount of toxic chemicals (e.g. NO\textsubscript{2}, SO\textsubscript{2}) and dusts, which are dangerous for human beings. At the same time, the burning of these fossil fuels can produce a large amount of carbon dioxide, which can lead to the greenhouse effect and hence climate change.

Solar energy is clean and plentiful, which make it a very good alternative, and it can be harvested by solar cells (SCs). However, despite the rapid progress of SC technology, there are still some major barriers that are holding back the large-scale implementation of solar energy, such as the difficulties in balancing the cost and efficiency of existing SCs.

Silicon-based SCs are currently dominating the solar energy market due to their low cost and long-term reliability. However, Si has an indirect band gap, which is a serious drawback for harvesting solar energy. So far, commercial Si SCs can only convert 8%–19% of the solar energy considering the affordable cost, which is far from efficient [2, 3]. On the other hand, most III–V materials have a direct band, which is beneficial for achieving high absorption with high material cost. In addition, the wide band gap range of III–V materials is beneficial for designing multi-junction SCs, which can make better use of the solar spectrum and maximise the energy conversion efficiency. Very high efficiencies (> 40%) can be obtained by integrating III–V multi-junction cells onto Ge SC substrates [4, 5]. However, the planar III–V materials and Ge
substrates needed for such devices are too rare and expensive for broad usage.

Innovations are thus needed to harvest solar energy with high efficiency and better economic viability. The ideal solution is to build high-efficiency III–V SCs on a low-cost mature Si platform and develop III–V/Si two-junction cells, which have been predicted to have a high efficiency of ~44% [6–10]. Nevertheless, there are large lattice and thermal expansion coefficient mismatches between III–V epi-layers and Si substrates, which significantly limits their integration [11, 12]. In addition, the growth of bi-atomic III–V materials on monatomic Si substrate suffers from the antiphase domain issue, which further hinders the integration. Thus, after 30 years of extensive research, there is still no solution on how to effectively implement this idea.

1.2. Advantages of nanowire SCs

In 1964, Ellis and Wagner presented a brand-new one-dimensional (1D) structure originally known as nanowhiskers [13]. This columnar-shaped structure later became known as nanowires (NWs) and has opened up a new world in material science. It offers a number of completely new features compared with the traditional structures. Lots of limitations in traditional material science can be solved with the help of this structure. For this reason, it has gained great attention [14–16]. The NW structure can facilitate material integration and provide more substrate choices. Due to the small diameter, the interface between the NW and its substrate is small, which is beneficial for strain relaxation. Even materials with large lattice and thermal expansion coefficient mismatches can be integrated. For example, Yoshimura et al reported that 90 nm thick InGaAs NWs can be grown on GaAs substrates without defects despite a large lattice mismatch of 2.1% between them [17]. For a number of years, NWs have been claimed to be a highly-promising structure to solve the thermal expansion mismatch and lattice mismatch issues in material integration, thus promoting the development of III–V/Si SCs.

When used for SCs, the NW structure can exhibit superior photon harvesting functions compared with thin-film structures. It has a 1D structure, subwavelength diameter and high refractive index, which make its electromagnetic modes leaky and more effective in interacting with the outside environment [18]. As a result, the absorption cross-section of the NW can be significantly enlarged compared with its physical size, which allows the NW to absorb light from outside the projected area [19] (figure 1(a)). In addition, the waveguide modes of the NW can guide the light to propagate along its axis, which can further enhance photon absorption. As a result, light absorption in the standing NW can be more than an order of magnitude more efficient compared with that predicted from the Lambert–Beer law [19]. Moreover, by controlling the NW diameter, it can control the optimum absorption wavelength of the leaky mode resonances (figure 1(b)) [20–23], which can offer wavelength-selective absorption enhancement. In addition, with some other geometry parameters [24], such as the NW length [25], profile [26] and density, tailoring the absorption spectrum is much easier compared with thin-film devices, which is beneficial for current matching in multi-junction devices [27].

Thin-film SCs without an antireflection coating exhibit severe light reflection at the surface and can lose between 10%–50% of the incident photons. Thus, thin-film SCs require fabrication of a highly complicated antireflection layer, which can result in high fabrication cost [28]. When a large number of NWs stand together on the substrate surface, they can form a refractive index transition layer, which can eliminate the abrupt refractive index change between the air and the device surface [29], and thus significantly reduce the light reflectance [30–32]. In addition, arrays formed by NWs with subwavelength dimensions can offer a strongly enhanced light scattering effect [33, 34]. Therefore, the light propagation path in the NW absorbing media can be significantly elongated (figure 1(c)), which makes its effective optical thickness exceed the Lambertian limit easily [35–38]. As demonstrated by Garnett et al [39], the light propagation path in their NW arrays is 73 times longer compared with the actual physical thickness of NW layers. With the NW structure, the intensity of transmitted light was greatly decreased due to the enhanced light trapping and absorption (figure 1(d)). This advanced light trapping ability can greatly enhance the light absorption within a wide range of incident angles [40, 41]. Diedenhofen et al reported that their InP NWs with a 3 µm length can have a broad band and omnidirectional absorbance in excess of 80% in the wave range from 400 nm to 850 nm [26]. Zhang et al showed that their off-substrate GaAs NWs, with a length of 1.6 µm and a low surface coverage of only 15%, exhibit a solar-weighted above-band-gap absorptance of 94% [23]. Wallentin et al reported that the short-circuit current generated by their InP NWs (~1.5 µm long, 12% surface coverage) is about 83% that of the best InP planar cells [42]. In addition to their enhanced absorption, the NW can also function as a waveguide to lead the below-band-gap photons into the substrate, which is highly beneficial for multi-junction cells. As demonstrated by Diedenhofen et al [26], their InP NWs can channel more than 90% of the below-band-gap photons into the underlying substrate. Due to the advanced photon harvesting functions, the NW SCs can employ a very small amount of material to achieve high absorption efficiency. Compared with the thin-film technique, this is beneficial for saving rare materials and greatly reducing the production costs.

Despite the great advantages, the efficiency of NW SCs is only about 15%–18%, which is still lower than that of traditional thin-film SCs. This is due to some limitations in knowledge and technologies that make device optimisation challenging, such as the high density of defects and unclear doping mechanism. In recent years, there have been significant breakthroughs in this knowledge and in these technologies, which set a strong foundation for the optimisation of NW SCs. Thus, this review summarises these breakthroughs as it will be beneficial for researchers to use all the updated knowledge and technologies to overcome the remaining challenges in making high-efficiency low-cost NW SCs.

The GaAsP NW is one of the most promising NWs for photovoltaics. It contains two group-V elements but only one
Figure 1. (a) External quantum efficiency of vertical and horizontal GaAs NW SCs normalised by indicated projected area. For the vertical standing SC, a 15-fold increase in photon collection was obtained close to the band gap. Inset is a scanning photocurrent measurement on the inset single vertical NW device for 800 nm excitation laser wavelengths, normalised to the incident photon flux. Reprinted by permission from Springer Nature [19]. Copyright (2013). (b) Absorptance as a function of wavelength and NW diameter. Adapted with permission from [21]. Copyright (2013) American Chemical Society. (c) Illustration of light trapping in NW arrays. (d) Transmission spectra of thin silicon window structures with and without NWs. There is a large intensity reduction and red shift in the transmitted light for the samples with NWs, suggesting strong light trapping. Adapted with permission from [39]. Copyright (2013) American Chemical Society.

group-III element, which can be easily grown by the self-catalysed mode [43]. The GaAsP is a ternary material with large band gap and lattice constant coverages, which can offer more freedom in band structure design and better lattice match compared with binary materials. Its band gap can cover operating wavelengths ranging from green (550 nm) to near-infrared (860 nm) at room temperature, which is highly useful for SCs. Especially, the two-junction tandem SC, consisting of a 1.7 eV GaAsP NW junction and a 1.1 eV Si junction, can have a high efficiency of ~44%. It needs to be noted that above a certain P content (~50%) the band gap will become indirect, limiting the efficiency of the SCs [44]. In this review, great attention has been given to the development of GaAsP NWs for making high-efficiency NW/Si tandem SCs. Dilute-nitride NWs like GaNAsP or GaNAs with similar element combination can also have similar properties [45, 46].

2. NW growth

The self-catalysed growth mode has a fast growth rate and can produce NWs with a low defect density, which has gained great attention. However, this growth mode is highly complicated due to the participation of the catalytic droplet, which makes the NW growth and SC design significantly different from those of traditional thin-film devices, and even different from NWs grown by other modes. Here, the new development and remaining challenges of critical techniques in the growth of self-catalysed NWs for SCs are reviewed using GaAs(P) NWs as an example.

2.1. Self-catalysed growth mode

The droplet-catalysed mode is also called VLS mode because it involves the vapour, liquid and solid phases [13]. This mode has significant advantages, such as fast growth rate and low defect density. The growth is assisted by the nano-scale catalytic particle that is normally made of low melting point metals. Depending on the catalyst particle material, this growth mode can be further divided into two categories, namely foreign-metal-catalysed and self-catalysed growth [47]. In foreign-metal-catalysed NW growth, a lot of metals can be used as the catalyst particle, in particular Au [47]. However, Au has a number of drawbacks, which make it rather unpopular. When incorporated into Si, it can form mid-gap energy states, which can affect
the device performance [48, 49]. For Au-catalysed NWs, the Au can be incorporated at a rather high concentration. For example, it has been found that the Au doping concentration in the GaAs and InAs can be in the order of $10^{17} - 10^{18} \text{ cm}^{-3}$ [48, 50]. Breuer et al. [49] made a direct comparison between Au- and self-catalysed GaAs NWs, and found that Au-catalysed NWs are lower in photoluminescence intensity and shorter in carrier lifetime. In addition, during the shell growth, the presence of the Au particle can seriously degrade the shell morphology. The NW tip and the sidewalls adopt VLS and VS growth mode, respectively [51], which lead to a different growth rate, composition profile and doping concentration [52].

To overcome the aforementioned problems, self-catalysed growth mode was developed. In this mode, the catalyst droplet is normally made of group-III metal with a low melting point and a high boiling temperature. Because the element of the catalyst droplet is also the same as one of the NW components, it does not introduce extra contamination. The substrate used for the growth needs to be covered by a masking layer to suppress 2D growth. The mask can be made of SiO$_2$ or SiNx. The mask requires some nano-sized pin holes to expose the substrate lattice (figure 2(a)). Before the NW growth, the catalyst droplet is formed inside the pinholes (figure 2(b)). Once the droplet is oversaturated by the group-V flux, the core NW growth starts (figure 2(c)). At the same time, the group-III flux should be kept open to avoid the droplet being consumed. When the core NW growth is finished, the catalytic droplet can be consumed by keeping only the group-V flux open (figure 2(d)). At this point, the VLS growth can be terminated, and then the shell growth begins (figure 2(e)). This growth mode is highly suitable for growing NW containing more than two group-V elements but prefers having only one group-III element. For example, GaAsP [53] and GaAsSb [54] NWs with a large composition coverage have been demonstrated.

### 2.2. Substrate preparation

To reduce the SC cost, it is preferable to use Si substrates. The self-catalysed NWs can be grown on both unpatterned and patterned Si substrates. NWs grown on unpatterned substrates are randomly positioned due to the self-assembled mechanism. Because the substrates do not need expensive treatment, this method can be used to produce low-cost NWs. To promote 1D NW growth, the Si surface needs to be covered by a layer of oxide that can be thermal oxide, native oxide or oxide made of hydrogen silsesquioxane (HSQ). On the oxide layer, it is necessary to form pinholes through which the Si lattice can be exposed and provide a nucleation template for NW growth. A thin thickness and a high roughness of the oxide layer are preferable for the formation of the pinholes. Depending on the surface roughness, the critical thickness is 1–2 nm for thermal and native oxides, and 5–6 nm for HSQ oxide [55].

The use of a pattern can help to improve NW uniformity, which is beneficial for achieving an almost identical performance for each NW. By controlling the pitch of the patterned holes, the distance between the NWs can be controlled, which can help to optimise optical absorption [56, 57]. In the case of multi-junction SCs, such as an NW/Si double-junction SC, this optical absorption adjustment is extremely useful because it can greatly facilitate current matching between junctions. Moreover, because material deposition on the mask layer is not energetically favourable, the parasitic bulk (cluster) growth which often accompanies unpatterned NW growth can be suppressed. This can significantly reduce photon losses by the absorption in the clusters and hence improve the SC efficiency. During SC device fabrication, the absence of clusters can facilitate material deposition/coating, making it much easier to achieve good uniformity. In addition, the insulating mask layer can also help to reduce the current leakage and hence improve the performance of the SCs.

A patterned mask can be made from SiO$_2$ or SiNx with nanoimprint or electron-beam lithography [58–61]. The thickness of the mask is commonly between 10 and 40 nm [62, 63]. The technique for growing NWs on patterned III–V substrates, such as GaAs NWs on GaAs substrates [64], InAs NWs on InAs substrates [22] and InP NWs on InP substrates [65], is well developed. However, due to the oxidation of the Si in the patterned holes, the growth on patterned Si was facing low-yield and low-repeatability issues (figure 3(a)) [66]. HF is commonly used for cleaning the Si surface, because the Si surface can be passivated by H-termination after the oxide is removed [67, 68]. However, this cleaning technique did not improve the growth quality on patterned Si, because the oxidation speed of Si is too fast once it is exposed to oxygen,
and the NW growth is very sensitive to the cleanness of the patterned holes.

Thus, an in situ high-temperature deoxidisation (HTD) step was developed to remove the oxide from the patterned holes [69]. Immediately prior to the NW growth, the substrate is heated to ~900 °C, which can thoroughly clean the patterned holes without damaging the mask. However, this causes another problem, when the excessively large nucleation area inside the patterned holes makes the group-III material consumption speed very fast, suppressing the formation of the catalytic droplet (figure 3(b)). A Ga pre-deposition (GPD) step was then developed to assist Ga droplet formation. When the substrate cooled down to the growth temperature after HTD, Ga is deposited first until the Ga droplet fills in the patterned holes. Then, the NW growth starts. With the HTD and GPD techniques, Zhang et al. for the first time demonstrated the growth of self-catalysed GaAsP NWs on patterned Si substrates with high yield and good repeatability (figure 3(c)) [69].

It should also be noted that for droplet-catalysed NW growth, the catalytic droplet size should not be bigger than the patterned hole size. If the hole size is larger than that of the catalytic droplet, there will be some oxide-free Si areas in the hole that are not covered by the droplet. Some of the Ga adatoms are deposited in this area through VS growth mode, which can reduce the Ga replenishment for the droplet and hence reduce the NW length. Moreover, this VS growth can also develop along the NW length, degrading the NW morphology.

2.3. Uniformity control

Good NW uniformity is critical for achieving high-performance devices uniformly across the device. Development of the uniformity control technique for self-catalysed NWs, especially on non-patterned substrates, is critical for producing high-performance and low-cost NW SCs. The growth of self-catalysed NWs is controlled by catalytic droplets that are in the nanoscale. At this small scale, the Gibbs–Thomson (GT) effect can greatly increase the balanced vapour pressure of the droplets [70, 71]. As a result, the smaller droplets have a lower effective supersaturation, which makes them grow at a lower rate and hence a shorter length (figures 4(a) and (b)). In addition, the GT effect can also affect the composition uniformity of NWs. For example, GaAsP NWs with a smaller diameter have a higher P content. During the growth, the vapour pressure of the droplets with a smaller size can be increased more compared with larger ones due to the GT effect [24, 71], which results in a stronger out-driving force to push out the group-V elements inside the droplet to lower the effective supersaturation. Therefore, the smaller-sized droplets have a higher barrier for incorporating new group-V atoms. Compared to As, P has a higher chemical potential with Ga, which gives it a stronger nucleation ability [72]. Therefore, the incorporation efficiency of P is less affected compared with As when the droplet size is decreased, making the smaller NWs richer in P.

The NW uniformity can be greatly improved by controlling the uniformity of the droplet size. As can be seen in figures 4(c) and (d), the GaAsP NWs with a good uniformity of 93.4% can be achieved using droplets with a highly uniform diameter. With the use of patterned Si substrates to further increase the uniformity of the droplet size, the NW uniformity can be further improved, which can be seen in figure 3(c).

It needs to be mentioned that there are some NWs grown in different directions than vertical. This could be caused by the 3D twinning which happens at the initial stages of NW growth on Si substrates [73]. Reducing the droplet size can suppress the nucleation of 3D twinning at the initial stages and hence suppress the multi-directional growth [73].

2.4. Composition control

One significant advantage of ternary material systems, compared with binary material systems, is the ability to adjust the energy band gap to cover a wide-operation wavelength range [69, 75, 76]. The NW composition control is therefore of critical importance. For example, when building GaAsP NW/Si tandem SCs, the band gap of GaAsP NWs has to be tailored to be ~1.7 eV to maximise the efficiency.

The differences in element incorporation in self-catalysed growth were investigated by growing GaAsP NWs with different P/(P + As) flux ratios (ranging from 7.2% to 50%). GaAsP NWs with P content ranging from 10% to 75% have been achieved. By investigating the correlation between the P percentage in the NW and the flux, the incorporation coefficient of P was found to be nearly three times (2.97) as large as that of As (figure 5(a)). This is in stark contrast to the thin-film planar epitaxial growth by both gas-source and solid-source
Figure 4. (a) Side view SEM image and (b) length-diameter summary of NWs grown with different droplet sizes. (c) Side view SEM image and (d) length summary histogram of GaAsP NWs grown with highly uniform droplet sizes. The inset in (d) is the diameter summary histograms of (c). Reproduced from [74]. CC BY 4.0.

Figure 5. (a) P/(P + As) ratio in the NW as a function of that in the flux. The thin-film GaAsP grown by gas-source (pink) and solid-source (blue) molecular beam epitaxy are also shown as a comparison. Illustration of (b) the thin-film VS and (c) the NW VLS growth modes. Reprinted with permission from [53]. Copyright (2013) American Chemical Society.

In thin-film growth, which follows the VS growth mode and is normally under group-V element-rich conditions, the growth rate is controlled by the group-III element. Group-V adatoms cannot be incorporated into the lattice if they are unable to meet the group-III adatoms and will desorb back into the vapour in the end. As adatoms have a larger sticking coefficient and hence longer surface lifetime and migration length, they have a larger possibility to meet Ga adatoms [79–81]. Therefore, the As adatoms are much easier to incorporate into the epitaxial layer compared with P, despite the fact that As has lower chemical activity [82, 83].

In the case of core NW growth, it is performed in the VLS mode, which is under a group-III element-rich environment. The growth is thus limited by the group V flux. The influence of P and As in Au-catalysed NWs was investigated theoretically by Glas [72]. According to that model, P was shown to have a stronger influence on the chemical potential of the Au droplet, and thus P can lead to a higher increase in the chemical potential for the same concentration. This suggests that the inclusion of As atoms in the droplet would be easier than that of P in GaAsP NW growth due to the smaller increase in the chemical potential for the same amount. However, according to the experimental results, the
extra chemical potential increase in the catalyst droplet does not form a barrier for absorbing P into it. Instead, the increased chemical potential caused by the absorption of P can enhance the nucleation rate of the GaP relative to that of GaAs. Besides, P has a much stronger chemical bond with Ga compared to that with As, which endows it with a stronger nucleation ability in the catalytic droplet.

From the analysis above, it can be seen that the incorporation ratio of P/As in thin-film growth is more governed by their physical properties, while that of core NW growth is more controlled by their chemical properties. It should be mentioned that NW shell growth follows the VS growth mode. Thus, this finding can be used to guide composition control for both core and shell NWs.

2.5. Crystal quality control

In contrast to thin-film crystals, for which each material can only select one crystal structure, either wurtzite (WZ) or zincblende (ZB) phase, the core NWs can adopt both WZ and ZB crystal structures in the same NW, forming stacking faults. These two phases have a type-II band alignment [84–88], which can separately confine the electrons and holes to reduce overlap of their wave functions. This can lead to a rapid quench of the direct-band-gap carrier recombination and cause a long carrier lifetime [89, 90], which could be beneficial for photovoltaics. Recently, Jansson et al. suggested that the high stacking fault density could cause severe non-radiative recombination, which in turn can limit the SC efficiency [91]. Moreover, the stacking faults can cause significant problems for NW shell growth. For example, the GaAsP NW is bounded by 110 facets that are low in surface energy [92]. The stacking faults can change the atom arrangement at the surface and create new facets with higher surface energies. The nucleation is therefore energetically more preferable at these sites, which causes non-uniform shell growth. This phenomenon is even more severe when growing p-type doped shells (figure 6(a)). Be is commonly used as the p-dopant and known to act as a surfactant that can improve the mobility of Ga adatoms [93]. A high concentration of Be doping flux can promote Ga to move to and accumulate at the stacking-fault locations. As a result, these locations grow much faster than the defect-free locations, leading to the formation of big lumps and generating new dislocations, e.g. high-angle grain boundaries [94]. The quality of the NW tip can also greatly influence the shell growth. After the core NW growth, the catalytic droplet needs to be consumed for the succeeding shell growth. For the non-optimised droplet consumption process, multiple nucleation sites can be activated, which results in a significantly enlarged defective tip with saw-tooth-like irregular top facets. During the shell growth, the saw-tooth-like top facets provide more energetically favourable nucleation sites compared to the smooth 110 sidewalls with a lower surface energy. Besides, the tip is more effective in collecting source materials, since the fluxes come from the top. Therefore, the tip can grow faster and develop into a large irregular lump with a high density of defects.

To achieve high-quality NWs, it is critical to eliminate the defects in the NW core, including the stacking faults in the NW body and the dislocations at the NW tip. Although the detailed nature of the stacking fault formation in droplet-catalysed growth is still under debate [95], it is widely accepted that the surface energy of the catalytic droplet has significant influence on this [96]. Be can effectively lower the surface energy, which can promote ZB nucleation and suppress WZ nucleation. In addition, the low surface energy is also beneficial for suppressing defect formation during the droplet consumption procedure. This has been demonstrated by droplet-consumed GaAsP core NWs that have achieved a defect-free crystal structure up to the top-most bi-layer (figure 6(b)). On the defect-free NW cores, high-quality GaAsP shells can be grown, producing core-shell NWs with a highly uniform diameter along the length and a highly regular tip bounded by well-developed facets (figure 6(c)). Because the core NWs are defect-free, the shell does not normally have any observable defects. Due to the elimination of the defects, these NWs with good core-shell quality can have more than 15 times enhancement in photoluminescence emission compared to defective ones.

2.6. Doping mechanism

Doping is important for semiconductors to change the electrical properties. However, the doping mechanism of self-catalysed NW is highly complex due to the participation of the liquid phase [13]. There have been long-term arguments about whether the dopants are incorporated through droplets, sidewall diffusion or sidewall growth [98, 99]. To clarify the doping path of self-catalysed NWs, atom-probe tomography (APT) measurements, with the advantages of quantitatively identifying the dopant distribution inside the NWs [100–102], were performed on GaAs NWs of various nominal Be doping concentrations (figures 7(a)–(c)). The measurement reveals that all the NWs have uniform Be dopant distribution along the entire length and radially across the whole diameter. There is no Be accumulation layer at the NW surface or segregation inside. If Be dopants were incorporated through sidewall diffusion or growth, the tip regions would be expected to have a lower doping concentration than the middle regions. Therefore, this observation explicitly suggests that dopant incorporation occurs predominantly through catalytic droplets (figure 7(d)).

Due to the high growth temperature, good thermodynamic equilibrium is set up at the growth interface between the liquid droplet and the solid NW, which is beneficial for the Be doping atoms to be incorporated at the lowest energy sites in the crystal lattice, namely Ga sites [103]. Even when the NW is saturated at a high doping flux, the majority of the dopants are still substitutional for Ga, forming effective doping. The excess Be atoms will be filtered out to accumulate in the catalytic droplet. This phenomenon is completely different from thin-film vapour–solid growth, where the segregation and precipitation of large amounts of doping material can lead to deterioration in film quality [104–108]. Of course, a too high Be
concentration inside the catalytic droplet in NW growth can lead to a large change of the droplet properties, such as lowering the surface energy, and cause the failure of the NW growth (figure 7(d)).

Si is commonly used as an n-type dopant in III–Vs. However, it has amphoteric behaviour, and its atoms can be incorporated as donors on Ga sites or as acceptors on As sites [109]. Due to this special doping path, and the Ga-rich environment, Si gets an even higher chance to replace As. Thus, self-catalysed NWs are commonly doped into p-type to avoid amphoteric behaviour of n-type doping.

The shell NW growth follows VS mode. The doping mechanism of thin-film growth should be applicable to the doping of shell NWs, but it needs to be modified to fit the growth conditions of the NWs, which is still missing.

2.7. P-i-n junctions

For SCs, the p-i-n junction is the core structure that can form built-in electric potential for separating and collecting photo-generated carriers (electrons and holes). The self-catalysed NW commonly adopts a radial structure that is also called the core–shell junction, and can provide an enhanced carrier extraction ability.

In SCs, the generated carriers are difficult to extract if they are more than one diffusion length away from the space charge region. Therefore, the minority carrier diffusion length determines the carrier collection efficiency. If the defect density increases, it will reduce the minority carrier diffusion length and hence the carrier collection efficiency [111, 112]. In thin-film structures or NWs with an axial junction (figures 8(a) and (b)), the carrier collection is along the same direction as the
Figure 7. (a)–(c) APT analysis of NWs with a Be doping density of $1.28 \times 10^{19}/\text{cm}^3$. (a) Be concentration profile along the NW length. Insets above are 3D tomographic reconstructions of each NW. (b), (c) Radial Be atom distribution maps of the 100 nm segments marked by the square in (a). (d) Illustration of doping mechanism in the self–catalysed NW growth. Reprinted with permission from [110]. Copyright (2018) American Chemical Society.

Figure 8. Illustration of the propagation path of photon and carriers in (a) thin–film SCs, and NW SCs with (b) an axial junction and (c) a radial junction.

solar photon travelling path [113, 114]. In order to adequately absorb the photons, a thick absorber is normally preferred. However, this puts a high demand on the crystal quality to give carriers a long enough diffusion length, which can significantly increase the cell cost. For NW SCs with a radial junction (figure 8(c)), the light is absorbed along the entire NW, while the carrier separation occurs along the radial direction. Due to the small radius, the radial distance carriers need to pass before being collected is smaller than or comparable with their diffusion length. Therefore, the decoupled light absorption and carrier separation paths can ensure a high quantum efficiency. In addition, the large surface-to-volume ratio of the NW gives it a large junction area, which can further increase the carrier separation efficiency. As a result, the use of lower-purity materials can still result in a high efficiency and hence greatly reduce the SC cost. Moreover, the radial junction has the active region far away from the surface, which is different from the axial junction with the active region exposed/close to the surface (figures 8(b) and (c)). Therefore, the radial junction is more resistant to surface recombination. It is predicted that an NW SC with a radial junction can have a high internal quantum efficiency of 95% even when the surface recombination velocity is as high as $10^5$ cm s$^{-1}$, while axial devices will require substantially lower values of $10^3$–$10^4$ cm s$^{-1}$ to achieve the same level of efficiency [115, 116]. It is also predicted that the open-circuit voltages of radial devices can be nearly twice as large as those of axial devices [115, 116].
However, at this time the fabrication of devices with a radial junction is rather challenging. So far, there is still a lack of systematic optimisation of the radial junction. Boulanger et al. [117] reported an off-axis electron holography technology to measure the electrostatic potential distribution across the radial p-i-n junction of GaAs NWs. This could provide a quick and cost-effective way to perform the optimisation work.

2.8. Surface passivation

The high-density surface states of NWs can act as efficient carrier-trapping centres, causing band bending and preventing the formation of Ohmic contact. They are also high-efficiency non-radiative recombination centres, and can consume a large number of photon-generated carriers, causing low device efficiency. NWs have very a small diameter and hence a large surface-to-volume ratio. The carriers inside can thus be captured by the surface states with much higher efficiency compared with thin-film devices. Therefore, it is highly important to passivate the NW surface.

Various surface passivation techniques have been developed. Chemical passivation, such as using sulphide solutions, can provide an effective instant passivation; however, it is lacking long-term stability [118, 119]. A more effective and robust passivation technique is using in situ-grown inorganic materials with a wider band gap that can block carriers from reaching the surface states. However, to achieve a good passivation effect, a thick passivation layer is preferred [120], which puts high requirements on the lattice match between the NW and the passivation layers. The widely used surface passivation materials for GaAs-based NWs include AlGaAs [120–123], AlInP [124] and InGaP [125]. With the exception of AlGaAs, lattice matching requires careful tuning of growth parameters for each individual system, which is often laborious and time-consuming in non-planar NW growth. The use of thick lattice-mismatched material will introduce a large strain and seriously degrade the crystal quality [126]. It can also inevitably affect the band structure and morphology of the NW, limiting further device performance.

A more widely-used passivation method for NW SCs is to reduce the surface recombination rate. InP material has a much smaller surface recombination rate (~170 cm s$^{-1}$) compared with other materials, such as GaAs (5.4 × 10$^5$ cm s$^{-1}$). This makes it highly suitable for passivating the NW surface and its effectiveness has been widely demonstrated. For example, by adding an InGaP surface passivation layer, Holm et al. increased the efficiency of GaAsP NW SCs from ~6% to > 10% [75]; Mariani et al. improved the efficiency of their GaAs NW SCs from 1.02% to 6.63% [127]; Nakai et al. also enhanced the efficiency of GaAs NW SCs from 0.71% to 4% [128].

3. NW Solar Cells

In terms of size, there are single and large-area NW SCs. Single-NW SCs can be used to study the intrinsic properties and limiting factors of NWs, and can provide valuable information to aid in designing and fabricating large-area SCs. They can also be used as highly compact detectors in integrated photonic systems. Large-area SCs can be used for solar energy harvesting or photonic signal detection. Thus, both of them have gained great attention and are developing rapidly. Here, important developments of the NW SCs are reviewed, regardless of the growth modes.

3.1. Single nanowire solar cells

To fabricate single-NW SCs, one of the major challenges is the difficulty to make high-quality Ohmic contact, especially p-type contact. NWs have a very large surface-to-volume ratio. The high-density surface states can easily deplete the NW surface layer or even the entire NW if the diameter is small by capturing a large number of carriers. This can cause surface energy band bending and surface Fermi level pinning to the middle of the band gap, causing the formation of the Schottky contact rather than the Ohmic contact. To achieve good p-type Ohmic contact, both the NW doping and the annealing have been shown to be critical. For example, by carefully controlling the doping level and the annealing condition, good AuZn Ohmic contacts can be created on p-GaAs NWs with a diameter as small as 50–60 nm [129]. The technique for achieving this small diameter will meet the fabrication requirements of most single-NW SCs or detector.

For single-NW SCs with a radial junction, fabrication is rather challenging. Due to the core–shell structure, one of the electrodes needs to make contact at the central/core region of the NW. Thus, shell removal is necessary. For example, for self-catalysed GaAsP NWs, the p-type doped core region is covered by i- and n-shells. To make p-type contact, it is necessary to remove the n- and i-shells, which is very difficult due to their small size. Shell removal is normally done by chemical wet etching, which is not sufficiently effective for uniform and precise shell removal. In particular, for NWs with stacking faults, the defective region has a higher surface energy and hence a faster shell etching rate. After etching, the remaining core NW has a non-uniform diameter with some heavily corroded regions, thus making it difficult to support fabricating high-quality contacts. In addition, each layer in the p-i-n junction is very thin and the contact metal diffusion during annealing can damage it. Therefore, most of the papers reported a low fill factor of < 65%. Only Holm et al. reported a good fill factor of 72%–77% (figure 9(a)) [75]. Zhang et al. developed a dry etching method using an ORION NanoFab focused ion beam (FIB), which can perform the shell removal uniformly and precisely [129]. With this method, they fabricated high-quality contacts to the NW and achieved a record-high fill factor of 80.5%.

For single-NW SCs with an axial junction, the fabrication process is greatly simplified. There is no etching process needed. In addition, the junction region is far away from the contacts, which gives more freedom in annealing the contacts. Barrigón et al. showed that GaAs p-n junctions can exhibit excellent diode characteristics with a high rectification ratio in excess of 10$^5$, and an ideality factor of ~2 [131]. Due to
the simplicity of the fabrication process, the SCs tend to have good fill factors. Li et al. fabricated InP axial p-i-n junction SCs and demonstrated a fill factor of 70% and an efficiency of 7.73% [132]. The axial junction has better tolerance to doping variation, which can lead to a high open-circuit voltage. For example, Yao et al. achieved an open-circuit voltage as high as 0.716 V from GaAs NWs with an axial p-i-n junction, which is higher than that reported for the best GaAs NW SCs with a radial junction (~0.6 V) [64]. Wallentin et al. also achieved an open-circuit voltage of 0.906 V from axial InP NW SCs which exceeds that of the planar counterparts [42]. The axial structure is also beneficial for building tandem junction devices. Heurlin et al. connected two axial p-i-n junctions by an axial tunnel junction inside a single-InP NW [130]. They achieved an open-circuit voltage of 1.15 V, which is an increase of 67% compared to a single-junction device (figure 9(b)). This can be used to build high-voltage detectors or large-area tandem SCs. To build a high-quality axial junction, it is important to suppress the short-circuit shunting paths and non-radiative recombination centres at the NW surface. Cui et al. eliminated the unintentional radial growth on the sidewall of their InP NWs with an axial p-n junction, which greatly increased the rectification ratio from 10² to 10⁷, and improved the ideality factor from 3.17 to 2.12 [133].

So far, the single-NW SCs laying horizontally on the substrate can reach an efficiency of ~10% (figure 9(a)). If the NW is standing vertically on substrates, it can greatly increase the efficiency by improving the broadband light absorption and the angle insensitivity. Ko et al. fabricated a vertically standing single-InP NW SC, which demonstrated an absorption enhancement at a factor of 200%–400% over a wide spectral range of 400–800 nm due to a novel tapered antenna effect [134]. This enabled them to achieve an efficiency of 19.6% despite a low fill factor of ~48%. Krostrup et al. fabricated vertically standing single-GaAs NW SCs and observed a greatly enhanced light absorption effect due to its light-concentrating property [19]. Light absorption in their standing NWs was enhanced by a factor of 10 to 70 with respect to the equivalent thin-film structure. Their device demonstrated a short-circuit current of 180 mA cm⁻², which is more than an order of magnitude higher than that predicted from the Lambert–Beer law [135]. As a result, a high efficiency of 40% was achieved despite the low fill factor of 52% (figure 9(c)).

3.2. Large-area SCs

To make large-area SCs, a top-bottom contact configuration is commonly used. One electrode contacts the under-facet of the heavily doped substrate. The top electrode should be transparent and in contact with the tip of the NW arrays. Due to the 1D structure, it is difficult to achieve good interconnection of contact material between the NWs, which can result in a high sheet resistance. Therefore, in between the NWs, it is necessary to create a layer of transparent polymer to planarise the NW array. Cyclotene has been suggested as a suitable polymer [136]. This can support the deposition of a planar front contact with a low sheet resistance. For the axial p-i-n structure, this polymer layer can also ensure that the top electrode only contacts the top contact region without touching the other two regions. For a radial p-i-n structure, this polymer layer can be used to suppress the current leakage through the substrate.

Before depositing the top contact onto the NWs, it is necessary to etch back the polymer to expose the tip of the NWs. The axial p-i-n structure is more resistant to this etching, because its junction/active region is far away from the tip. In addition, annealing is necessary to improve the contact quality, but can cause metal diffusion which can damage the device if it diffuses into the junction area. The axial structure is more resistant to the contact metal diffusion. An efficiency over 10% has been demonstrated for large-area SCs with a single axial junction. In particular, Lberg et al. achieved 15.3% efficiency using bottom-up-grown GaAs NW arrays (figure 10(a)) [137]. Dam et al. demonstrated an efficiency of 17.8% with top-down-etched InP NWs (figure 10(b)) [138]. For SCs with a radial junction, the junction conformally covers the NW and is only a few hundred nanometres away from the surface. Both the polymer back-etching process and the contact metal annealing process can easily cause damage to the junction. As a result, this type of SCs commonly have a low efficiency of < 7.5%.
To reduce the cost and/or improve the efficiency, it is necessary to put III–V NW SCs onto Si substrates. In particular, if a 1.7 eV NW SC is placed onto a Si SC, this two-junction tandem SC can have a high theoretical efficiency of 44%. However, the NW SCs grown on Si commonly have a low efficiency, which could be due to the challenges in achieving good interface properties between the NW and the Si substrate. Yao et al made a significant breakthrough on this and developed a low-resistance connecting junction using a heavily doped $n^+$-GaAs/$p^+$-Si heterojunction [139]. Then, they made a GaAs NW/Si tandem SC and demonstrated an open-circuit voltage of 0.956 V and an efficiency of 11.4% (figure 10(c)). This is however still much lower than the single-junction NW SCs grown on III–V substrates.

Hybrid SCs, made by integrating inorganic NWs with organic polymers, can greatly reduce the cost. The organic polymers have low fabrication cost, but their photon-generated carriers are commonly low in mobility and short in lifetime...
[143], while inorganic semiconductor materials have a good carrier mobility and long carrier lifetime. Combining the advantages of the two material systems is promising for the fabrication of hybrid SCs with a good efficiency at a low cost. With the NW structure, the large surface area can greatly enhance the charge separation and collection efficiency [144]. Shankar et al demonstrated a four-fold increase in quantum yield by using TiO2 NWs to produce efficient carrier transfer properties [145]. The integration between poly(3,4-ethylenedioxythiophene):poly(styrene sulfonate) (PEDOT:PSS) and GaAs NWs can also greatly improve the SC efficiency from 0.29% to 5.8% [146]. To achieve good hybridisation, three factors are essential. First, it is important that the thickness of the polymer layer is smaller than the carrier diffusion/collection length. Tsai et al found from their Si NW arrays/3-hexylthiophene (P3HT) hybrid SCs that NWs with suitable polymer thickness can improve the short-circuit current by 61% and increase the efficiency from 0.96% to 2.31% as compared to that with very thick polymer layers [147]. Second, the interface quality is important for the integration quality. Yu et al introduced an intermediate 1,1-bis[(di-4-tolylamino)phenyl]cyclohexane (TAPC) layer into Si NW/PEDOT:PSS hybrid SCs [148]. The TAPC improved the morphology of PEDOT:PSS laying atop Si NWs, and suppressed the interface recombination and dark saturation current, which increased the open-circuit voltage and fill factor, and resulted in a high efficiency of 13.01%. Wang et al studied Si NW/PEDOT:PSS hybrid SCs and found that careful Si NW surface cleaning (for removing oxide and Au contamination) can improve the efficiency from about 10% to 12.4% [149]. Third, it is also important to engineer the energy levels in the organic layer to optimise the charge transfer at the hybrid interface [150]. So far, the efficiency of the hybrid SCs can be well over 10% [151]. Um et al even achieved an efficiency of 16.1% using Si NW/PEDOT:PSS hybridisation (figure 10(d)) [140]. However, these hybrid SCs with a high efficiency all use top-down-etched NWs. The reports on using bottom-up NWs still quote an efficiency lower than 5%, which could be hindered by the low carrier transport efficiency between the NW and substrate.

The Si substrate is a high-purity crystal, which is still rather costly for building SCs. Replacing the Si substrate with something cheaper would make SCs more cost effective. NWs with an outstanding strain accommodation ability have a much wider substrate selection compared with thin films. They are not limited to single-crystalline and polycrystalline substrates, and can also be grown on carbon nanotube, graphene, amorphous Si, fibre-textured silicon thin film, iodium tin oxide and glass [112, 152–156]. O’Donnell et al used glass substrates to grow Si NWs and made SCs with an efficiency of 5.6% [157]. Mohseni et al grew InGaAs NWs on graphene films and demonstrated SCs with an efficiency of 2.51% (figure 10(e)) [141]. A wide range of materials can be used for NW growth through surface treatment by graphene, which provides a highly promising method to greatly reduce the cost. Fan et al even demonstrated the direct growth of highly regular, single-crystalline n-Cds nanopillar arrays on aluminium substrates [158]. Then, they made SCs and achieved an efficiency of ~6%. Owing to the use of Al substrates, their SCs also show good flexibility and thus confirmed that this is a good method to make flexible SCs.

Developing NW-only SCs without using substrate can reduce the cost further. Due to the unique 1D structure, NWs can be easily peeled off in large scale from the substrate after being buried into polymers [159]. The NWs can then be used for making SCs and the substrate can be re-used for the next NW growth. SCs fabricated in this way also have the advantages of being lightweight and having good flexibility, which can have some special applications, such as integrating flexible cells into clothing. This type of SC has attracted widespread attention and is developing rapidly. For example, Han et al fabricated flexible SCs with GaAs NWs lying horizontally and achieved a high efficiency of 16% (figure 10(f)) [142].

4. Remaining challenges

Despite the great achievements in making NW SCs, there are still some remaining challenges:

(a) Most SCs with radial junctions still show a low open-circuit voltage. Each layer in the junction is very thin, which makes it very important but difficult to balance the effective dopant concentration in the p- and n-regions. In addition, both the ionisation energy of dopants and surface states can have great impact on carrier levels and hence the junction quality. Thus, systematic work is needed in the optimisation of p-i-n junctions. The off-axis electron holography technology proposed by Boulanger et al [117] could provide a quick and cost-effective way to perform optimisation.

(b) The top contact quality of large-area SCs still needs to be optimised to further improve the transparency, reduce the contact resistance, and achieve good interconnection of contact material between the NWs. Due to the filling of polymers in between NWs, the contact annealing process needs to be optimised to achieve good contact without damaging the polymer. For SCs with a radial junction, a repeatable polymer etch-back technique has to be developed to efficiently expose the NW tip without damaging the junction.

(c) So far, with the exception of theoretical modelling, systematic work is still lacking in optimising the NW diameter, period and length to maximise optical absorption. In addition, there is still a lack of work in optimising the current matching for tandem SCs.

(d) Despite substantial progress in developing surface passivation techniques, quantitative assessment of the passivation effect with high accuracy is needed to understand how much more optimisation is required. This can promote the development of new highly efficient and robust surface passivation techniques.

(e) There are lots of new breakthroughs in the growth of NWs and fabrication of NW SCs. These technologies need to be put together to optimise the performance of NW SCs.
For example, defect-free core–shell NWs can be used to build radial junction SCs free of the long-standing defective tip problem [97]. NWs grown on patterned Si substrates with good interface quality can be employed to improve the performance of Si-based III–V NW SCs, especially III–V/Si tandem SCs [69].

5. Conclusions and perspectives

Solar energy will be one of our major energy sources in the future because it is clean and renewable. Traditional SC technologies are facing challenges in balancing cost and efficiency. The NW structure is highly suitable for making high-efficiency and low-cost SCs, and has attracted great attention. For example, self-catalysed NWs are being widely used for SCs, but due to this field being new and immature, little is known about this growth mode, which poses big challenges for SC design, growth and fabrication.

Over the last few years, significant effort has been applied to understand and control this growth mode, which provided valuable information on the design and growth of NW SCs. Self-catalysed high-quality core–shell NWs can be grown on both patterned and unpatterned Si substrates with a high level of uniformity, which provides a solid foundation for III–V/Si integration. The core–shell NWs can have a pure zincblende crystal structure without significant defects. Both the composition control mechanism and doping mechanism have been thoroughly investigated, which respectively provides a theory to guide the adjustment of NW operation wavelengths and electrical properties. There is some preliminary work on building radial junctions. Fairly effective surface passivation techniques have been also developed.

In addition, there is also big progress in fabricating NW SCs. Single NWs have achieved an efficiency of ~10% and > 40% in horizontal and vertical structures, respectively. Large-area SCs have realised an efficiency of 15.3% using bottom-up-grown GaAs NW arrays, and 17.8% with top-down-etched InP NWs. There are also some initial trials in making NW SCs with more advanced structures. NW/Si tandem SCs have reached an efficiency of 11.4% using GaAs NWs; organic polymer/inorganic NW hybrid SCs have demonstrated an efficiency of >16%; NW SCs grown on cheap non-crystal substrates are capable of achieving an efficiency of ~6%; substrate-free NW SCs have also been fabricated and shown an efficiency of 16%.

Despite the great achievements in both NW SC growth and fabrication, there are still some remaining challenges. Significant work is still needed in optimising p–i–n junctions, top contacts, NW/Si interface quality, surface passivation techniques and NW dimensions to maximise optical absorption and current matching.

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References

[1] Tükel A, Karakus Y and Kahriman A 2009 Importance of thorium as alternative energy source and role of Turkey Int. Multi. Sci. Geo. Conf. SGEM: Surv. Geol. Min. Ecol. Manage. vol 2 p 429
[2] Almansouri I, Ho-Baillie A, Bremner S P and Green M A 2015 IEEE J. Photovolt. 5 968
[3] Essig S et al 2017 Nat. Energy 2 17144
[4] Abdul Hadi S, Fitzgerald E A, Griffiths S and Nayfeh A 2018 J. Renew. Sustain. Energy 10 015905
[5] Green M A, Emery K, Hishikawa Y, Warta W and Dunlop E D 2016 Prog. Photovolt. Res. Appl. 7 905–13
[6] Grassman T J, Carlin A M, Grandal J, Ratcliff C, Yang L, Mills M J and Ringel S A 2012 February spectrum-optimized Si–based III–V multijunction photovoltaics Proc. SPIE 8256 82560R
[7] Marti A and Araujo G L 1996 Sol. Energy Mater. Sol. Cells 43 203
[8] De Vos A 1980 J. Phys. D: Appl. Phys. 13 839
[9] Abdul Hadi S, Fitzgerald E A and Nayfeh A 2016 J. Appl. Phys. 119 073104
[10] Kurtz S R, Faine P and Olson J M 1990 J. Appl. Phys. 68 1890
[11] Shi J Y, Yu L P, Wang Y Z, Zhang G Y and Zhang H 2002 Appl. Phys. Lett. 80 2293–5
[12] Tang M, Chen S, Wu J, Jiang Q, Dorogan V G, Benamara M, Mazur Y I, Salamno G J, Seeds A and Liu H 2014 Opt. Express 22 11528–35
[13] Wagner R S and Ellis W C 2016 Appl. Phys. Lett. 4 89–90
[14] Lieber C M and Wang Z L 2007 MRS Bull. 32 99–108
[15] Yan R, Gargas D and Yang P 2009 Nat. Photon. 3 569–76
[16] Yang P, Yan R and Fardy M 2010 Nano Lett. 10 1529–36
[17] Yoshimura M, Tomioka K, Hiruma K, Hara S, Motohisa J and Fukui T 2011 J. Cryst. Growth 315 148–51
[18] Cao L, White J S, Park J-S, Schuller J A, Clemens B M and Brongersma M L 2009 Nat. Mater. 8 643–7
[19] Kroonstra P, Jørgensen H I, Heiss M, Demichel O, Holm J V, Aagesen M, Nygaard J and I Morral A F 2013 Nat. Photon. 7 306–10
[20] Hu Y, Li M, He J and LaPierre R 2013 Nanotechnology 24 065402
[21] Svensson J, Anttu N, Vainorius N, Borg B M and Wernersson L E 2013 Nano Lett. 13 1380–5
[22] Anttu N, Lehmann S, Storm K, Dick K A, Samuelson L, Wu P M and Pistol M-E 2014 Nano Lett. 14 5650–5
[23] Zhang J, Dhindsa N, Chia A, Boulanger J, Khodadad I, Saini S and LaPierre R 2014 Appl. Phys. Lett. 105 123113
[24] Wu P M, Anttu N, Xu H, Samuelson L and Pistol M-E 2012 Nano Lett. 12 1990–5
[25] Dai Y-A, Chang H-C, Lai K-Y, Lin C-A, Chung R-J, Lin G-R and He J-H 2010 J. Mater. Chem. 20 10924–30
[26] Diedenhofen S L, Janssen O T, Grzela G, Bakkers E P and Gómez Rivas J 2011 ACS Nano 5 2316–23
[27] Huang N, Lin C and Povinelli M L 2012 J. Appl. Phys. 112 064321
[28] Zhu J, Yu Z, Fan S and Cui Y 2010 Mater. Sci. Eng. R 70 330
[29] Hu L and Chen G 2007 Nano Lett. 7 3249–52
[30] Zhu J, Hsu C-M, Yu Z, Fan S and Cui Y 2009 Nano Lett. 10 1979–84
[31] Rayleigh L 1879 Proc. London Math. Soc. 1 51–6
[32] Muskens O L, Rivas J G, Algra R E, Bakkers E P and Lagendijk A 2008 Nano Lett. 8 2638–42
[33] Muskens O L, Diedenhofen S L, Kaas B C, Algra R E, Bakkers E P, Gómez Rivas J and Lagendijk A 2009 Nano Lett. 9 930–4
[34] Strudley T, Zehender T, Blejane C, Bakkers E P and Muskens O L 2013 Nat. Photon. 7 413–8
[35] Callahan D M, Munday J N and Atwater H A 2012 Nano Lett. 12 214–8
[36] Saeta P N, Ferry Y E, Pacifici D, Munday J N and Atwater H A 2009 Opt. Express 17 20975–90
[37] Yablonsovich E 1982 J. Opt. Soc. Am. 72 899–907
[38] Yablonsovich E and Cody G D 1982 IEEE Trans. Electron Devices 29 300–5
[39] Garnett E and Yang P 2010 Nano Lett. 10 1082–7
[40] Zhu J, Yu Z, Fan S and Cui Y 2010 Mater. Sci. Eng. R 70 330–40
[41] Wen L, Zhao Z, Li X, Shen Y, Guo H and Wang Y 2011 Appl. Phys. Lett. 99 143116
[42] Wallent J, Anttu N, Asoli D, Huffman M, Léger I, Magnuson M H, Siefer G, Fuss-Kauliweit P, Dimroth F and Witzigmann B 2013 Science 339 1057–60
[43] Colombo C, Spirkoska D, Frimmer M, Abstreiter G and I H 2011 Nano Lett. 11 1276–9
[44] Perez E D, Allen J E, May S J, Wessels B W, Seidman D N and Lauhon L J 2006 Nano Lett. 6 181–5
[45] Mohseni P K and LaPierre R R 2009 Nano Technology 20 025610
[46] Ahtapodov L, Todorovic J, Olk P, Mjåland T, Slåttnes P, Uccelli E, Arbiol J, Magen C, Krogstrup P, Russo-Averchi E, Givargizov E I and Matsushima Y and Gonda S-I 1976 Proc. London Math. Soc. 31 25–6
[47] Lberg I,´ı Lopez N, Yu K M, Tanaka T and Wulakiewicz W 2016 Adv. Energy Mater. 6 1501820
[48] Dick K A and Caroff P 2014 Nanoscale 6 3006–21
[49] Bar-Sadan M, Barthel J, Shtrikman H and Houben L 2012 Nano Lett. 12 2352–6
[50] Breuer S, Prißer C, Fliessikowski T, Brandt O, Grahn H T, Heiss M, Russo-Averchi E, Demichel O, Alarcon-Llad E and I Morral A F 2008 Nano Lett. 8 214–8
[51] Mattinei F, Tüünicüüoglu G, Rüffer D, Alarcón–Llador E and I Morral A F 2014 J. Cryst. Growth 404 246–55
[52] Kelzenberg M D, Boettcher S W, Petykiewicz J A, Turner-Evans D B, Putnam M C, Warren E L, Spurgeon J M, Briggs R M, Lewis N S and Atwater H A 2010 Nat. Mater. 9 239
[53] Lin C and Povinelli M L 2009 Opt. Express 17 19371–3
[54] Sun Y, Wang J, Hsu C-M, Algra R E, Bakkers E P and Lagendijk A 2009 Nano Lett. 9 930–4
[55] Baur B, Rudolph A, Soda M, I Morral A F, Zweck J, Schuh D and Reiger E 2010 Nanotechnology 21 435601
[56] Dick K A 2008 Prog. Cryst. Growth Charact. Mater. 54 138–73
[57] Plissard S, Larrieu G, Wallart X and Caroff P 2011 Nanotechnology 22 275602
[58] Heurlin M, Hultin O, Storm K, Lindgren D, Borgstrom M T and Samuelson L 2014 Nano Lett. 14 749–53
[59] Heurlin M, Stankeviči T, Mickevičius S, Yngman S, Lundgren D, Mikkelsen A and Samuelson L 2015 Nano Lett. 15 2462–7
[60] Heiss M, Russo-Averchi E, Dalmal-Mallorquí A, Tüünicüüoglu G, Matteini F, Rüffer D, Conesa-Boj S, Demichel O, Alarcon-Llador E and I Morral A F 2014 Nanotechnology 25 014015
[61] Higashi G S, Chabal Y J, Trucks G W and Raghavachari K 1990 Appl. Phys. Lett. 56 656
[62] Deppert K, Wallenberg L R and Samuelson L 2010 Nano Lett. 10 2214–18
[63] Muskens O L, Diedenhofen S L, Kaas B C, Algra R E, Bakkers E P, Gómez Rivas J and Lagendijk A 2009 Nano Lett. 9 930–4
[64] Yuan C, Povinelli M L, Dapkus P D and Zhou C 2014 Nano Lett. 14 3293–303
[65] Deppert K, Wallenberg L R and Samuelson L 2010 Nano Lett. 10 2214–18
