Enhancement of the Performance of GaP Solar Cells by Embedded In(N)P Quantum Dots

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Abstract

Improving the utilization of solar spectra of wide bandgap semiconductors that can potentially provide enough free energy is one of the promising strategies for realizing efficient and spontaneous integrated conversion of solar energy to chemical fuels. We demonstrate herein nitrogen doped InP quantum dots (QDs) embedded in wide bandgap GaP could improve the solar energy conversion performance. Photoelectrochemical experiments in contact with a nonaqueous, reversible redox couple indicated that the QD-embedded devices exhibited improved performance relative to devices without QDs, with short-circuit current densities increasing from 0.16 mA cm^{-2} for GaP-only devices to 0.23 and 0.29 mA cm^{-2} for InP and InNP QD-embedded devices, respectively. Additionally, the open-circuit voltages increased from 0.95 V for GaP-only devices to 1.11 and 1.14 V for InP and InNP QD-embedded devices, respectively, and the external quantum yield of the devices was also enhanced by the embedded QDs. The improvement is attributed to the absorption of sub-bandgap photons by the In(N)P QDs.

Graphical Abstract

Keywords

quantum dots, InNP, solar cells, indirect bandgap engineering, MBE

Introduction

GaP is not commonly used as the base absorber in terrestrial photovoltaic devices due to a bandgap ($E_g = 2.26$ eV) that is too large to absorb much of the solar spectrum, and because the indirect gap requires use of a relatively thick layer of GaP (> 5 μm) for effective absorption of light, relative to Group III-V semiconductors with direct bandgap. Nevertheless, the large bandgap and the electron affinity level (3.8 eV) of GaP may prove useful in an integrated solar-fuels device. Although the bandgap of GaP is considerably higher than optimal for an efficient solar-driven water-splitting device,¹ a tandem cell comprised of an n-GaP photoanode and a photocathode such as p-Si², p-WSe₂³, or p-Cu₂O^{4,5} may be useful for a solar-fuels device that couples the reduction of CO₂ to a fuel with the oxidation of water to O₂(g), because such multi-electron chemical reduction processes may involve larger catalytic overpotentials, and thus require higher overall photovoltages, than water-splitting devices. In addition, mass-transport limitations related to the limited concentration of CO₂ in the atmosphere may constrain the current density attainable for a CO₂-reducing device to values compatible with those obtainable from light absorbers with larger bandgaps than are optimal for solar-driven water splitting. Furthermore, GaP has a lattice constant very similar to that of Si (5.451 Å and 5.431 Å, respectively)⁶, allowing the growth on Si of high-quality GaP without anti-phase domains, through a buffer layer < 200 nm thick. Optoelectronic devices have in fact been fabricated using such a platform.⁶⁻⁹ Thus, a monolithic tandem cell comprised of a catalyst-loaded Si photocathode and an n-GaP photoanode, protected against corrosion and loaded with an oxygen-evolution catalyst,¹⁰ could provide an implementation of a functioning solar-fuels device. Techniques that enhance the absorption of GaP while reducing the thickness required for effective absorption of light, such as bandgap engineering^{11, 12} and nanostructuring¹³, will increase the utility of GaP not only for photoelectrochemical applications, but also for photovoltaic and optoelectronic applications.

One technique for tailoring the optoelectronic properties of GaP-based devices is embedment of quantum dots (QDs) into the GaP matrix. Self-assembled InP QDs can be fabricated on GaP substrates via the coherent-island Stranski-Krastanov growth mode¹⁴,

and a light-emitting diode (LED) with an emission energy of ~ 2 eV has been fabricated using this materials system¹⁵. The InP/GaP QD system has a band alignment intermediate between types I and II¹⁴, allowing further tailoring of the optoelectronic properties of the assembly through the incorporation of N into the QDs, to form dilute-nitride InNP QDs¹⁶.

Herein we report the performance of a p-i-n GaP device that contains embedded InP and InNP QDs. The QDs were fabricated on n-GaP(100) substrates ($n \sim 1 \times 10^{18} \text{ cm}^{-3}$) via the self-assembly process using gas-source molecular-beam epitaxy (GSMBE). Embedment of InP and InNP QDs into GaP p-i-n cells increased the absorption and enhanced both the short-circuit current density (J_{sc}) and the open-circuit voltage (V_{oc}) of the resulting devices, relative to structures that did not have embedded QD's in the material.

Experimental

Gas-Source Molecular Beam Epitaxy

A Varian Gen-II MBE system, modified to thermally crack gas-phase PH_3 to P_2 and H_2 , and GaP(001) substrates, were used to fabricate all samples described in this work. Elemental Ga and In, solid Si, solid Be, and N radicals produced using radio-frequency plasma activation were used as Group III sources, n-type dopant, p-type dopant and N source, respectively. The native oxide was removed from semi-insulating GaP substrates using a P_2 overpressure at $\sim 580^\circ\text{C}$. For single-layer In(N)P QD samples prepared for optical-quality studies, a 200 nm thick GaP buffer layer was grown at a rate of $1 \mu\text{m h}^{-1}$ after desorption of the oxide layer. The growth was then interrupted and the substrate temperature was lowered to $440\text{--}480^\circ\text{C}$. The temperature of the In Knudsen cell was raised until a deposition rate of $0.1 \text{ monolayer s}^{-1}$ (ML s^{-1}), as calibrated by reflection high-energy electron-diffraction (RHEED) intensity oscillation, was reached¹⁷. For InNP QDs, the N plasma was activated and set to a plasma power of 280 W using a N_2 flow rate of 1.2 sccm before the InNP QD formation process. The InP (InNP) QD self-assembly process began when the In (In and N) shutters were opened. When the InP (InNP) wetting layer exceeded the critical thickness, the InP (InNP) QDs self-assembled under a large compressive strain. The formation of QDs was confirmed by a change in the in-situ RHEED pattern from a streaky pattern to a chevron pattern, and was also

confirmed by atomic-force microscopy (AFM) images of the uncapped QD samples. Two ML of InP (InNP) were deposited in total for all samples in this study. After deposition of the QDs, a 20 nm GaP cap layer was grown at the temperature used for formation of the QDs to maintain the QD morphology and prevent indium desorption (due to its higher vapor pressure relative to gallium) during subsequent temperature increase. Note that all of the QDs in this study are smaller than 7 nm in height (figure 2) and no QDs should be exposed beyond the 20 nm GaP cap layer, which is confirmed in situ by the recovery of the streaky RHEED pattern during growth. An additional GaP cap layer, 30 nm in thickness, was grown at 580 °C on samples prepared for photoluminescence (PL) studies. If no cap layer were grown on top of the QDs, the QDs would oxidize once they are taken out of the vacuum chamber. This oxidation process degrades the quality of the QDs and significantly reduces the PL intensity. The photo-excited carriers would also tend to recombine through the surface states (mostly from the dangling bonds) on the QDs through the Shockley-Read-Hall mechanism. This non-radiative recombination would quench the PL signal intensity. Moreover, the uncapped QDs see a high surface potential (atmosphere) which can result in a shift of the quantum confined energy level and hence a shift of the PL peak position¹⁸. Therefore, a 50 nm-thick capping layer was used throughout the optical and device study to prevent device oxidation and degradation and to maintain the optical response consistency. For the In(N)P QD-embedded GaP solar-cell devices, a 200 nm thick GaP layer with a dopant concentration of $n=1\times10^{18} \text{ cm}^{-3}$ and an undoped 50 nm thick GaP layer were grown after desorption of the native oxide from the GaP(100) substrates ($n \sim 1\times10^{18} \text{ cm}^{-3}$) at $\sim 580 \text{ }^{\circ}\text{C}$, but before the growth of eight periods of In(N)P QDs. Each period of the QD structure consisted of a 2 nm thick delta-doped layer $\sim 1\times10^{11} \text{ cm}^{-2}$, an In(N)P QD layer grown at 460 °C, a 20 nm thick GaP cap layer grown at 460 °C, and a 30 nm thick GaP cap layer grown at 580 °C (to smooth the surface for the subsequent period of QD growth). An undoped 50 nm thick GaP spacer layer was grown after the final period of QD growth, followed by a 100 nm thick p-GaP layer ($p=1\times10^{18} \text{ cm}^{-3}$), and a p^+ -GaP contact layer ($p=3\times10^{18} \text{ cm}^{-3}$) that was 50 nm in thickness. An identical GaP p-i-n structure without any embedded QDs was also grown for comparison purposes. Previous studies showed that rapid thermal annealing (RTA) improves the PL intensity by improving the optical quality of samples,¹⁷ so the capped

QD samples and QD devices used in this study were covered by a GaP wafer (to protect the surface) and were annealed at 800 °C for 30 s prior to characterization.

Atomic Force Microscopy and Photoluminescence Measurements

Ex situ atomic-force microscopy (AFM) studies were performed under a vacuum pressure of $\sim 10^{-3}$ Pa using the dynamic force microscopy mode (SII Nanotechnology, E-SWEEP) with a Si cantilever.

For photoluminescence (PL) measurements, the samples were cooled to 4K in an evacuated cryostat. A continuous-wave 532 nm diode laser was used as an excitation source and the power density was set to $\sim 1 \text{ W cm}^{-2}$. The PL was dispersed by a monochromator (Princeton Instruments, SpectraPro 2300i) and detected by a cooled Si-CCD (Princeton Instruments, PIXIS 256).

Fabrication of Electrodes and Electrochemistry

The contact fabrication, chemical purification, and measurement methods have been described previously¹⁹. The substrate with In as ohmic contact was attached to a coiled, tin-plated Cu wire using conductive Ag paint, and epoxy (Hysol 9460F) was used to seal, face-down, the assembly at the end of a segment of glass tubing. The active area of each electrode was determined by the epoxy and measured using an optical scanner, with a typical value of 0.1-0.2 cm². Electrodes were left at room temperature to dry overnight. Before loading into the glovebox for photoelectrochemical measurement, the photoelectrodes were etched in a solution of 0.04% (v/v) Br₂ in methanol (CH₃OH, low water, J. T. Baker) for 15 s, followed by immersion in 1.0 M KOH(aq) for 30 s. The electrode was thoroughly rinsed with water and CH₃OH, then dried in a flow of N₂(g), and quickly transferred into a glovebox that was continually purged with ultra-high purity (UHP) Ar. Electrochemical data were obtained using a Biologic SP200 potentiostat. The nonaqueous electrochemical solutions each contained 1.0 M LiClO₄, 0.5 mM ferrocenium, and 90 mM ferrocene. The electrolyte solution was rapidly stirred with a small, Teflon-covered stir bar. Illumination was provided with an ELH-type tungsten-halogen lamp. The light intensity was determined by the use of a Si photodiode (Thorlab) that was previously calibrated relative to a secondary standard photodetector that was NIST-traceable and calibrated at 100 mW cm⁻² of Air Mass (AM) 1.5 illumination.

Electrodes were washed with CH₃CN prior to transfer to the electrolyte solutions. The working photoelectrodes were oriented vertically facing downward toward the incident light, with the light intensity adjusted by use of neutral density filters. To obtain the current density vs. potential (J–E) data, the potential was swept at a rate of 40 mV s⁻¹ in a three-electrode setup using a Pt mesh and a Pt wire as the counter electrode and reference electrode, respectively. External quantum yield measurements used the same electrochemical setup as that used for the collection of the J–E data. The J–E data were corrected for the solution resistance loss calculated from extrapolation of the linear portion of the J–E data measured using Pt disc electrode as a working electrode. The uncompensated resistance was measured to be 0.0057 ohm-cm².

Results and Discussion

Figure 1 presents a schematic illustration of the structures of the GaP and QD-embedded GaP devices that were fabricated for photoelectrochemical studies. Eight periods of In(N)P QDs were grown under high compressive strain within the undoped GaP layer of a GaP-based p-i-n structure. Optical-quality samples that did not contain either the p-GaP buffer ($p = 1 \times 10^{18} \text{ cm}^{-3}$) layer or the p⁺-GaP contact layer, and that contained only a single layer of In(N)P QDs, were used for photoluminescence studies. The formation of QDs was confirmed by a change in the in-situ reflection high-energy electron-diffraction (RHEED) pattern from a streaky pattern to a chevron pattern, and by atomic-force microscopy (AFM) images of the uncapped QD samples (Figure 2).

The AFM images were also used to calculate the densities and heights of the InP and InNP QDs. The densities of the InP and InNP QDs were $174 \pm 20 \text{ } \mu\text{m}^{-2}$ and $230 \pm 34 \text{ } \mu\text{m}^{-2}$, respectively. The heights of the InP and InNP QDs were $4.5 \pm 1.1 \text{ nm}$ and $5.8 \pm 1.3 \text{ nm}$, respectively. The presence of a N plasma during the QD assembly process did not affect the morphology significantly, although aggregated clusters with diameters of $\sim 100 \text{ nm}$ and heights of $\sim 30 \text{ nm}$ were found more frequently on InNP samples than on InP samples, possibly due to an increased rate of nucleation when the reactive N plasma was present.

Although such clusters are too large to contribute to the confinement energy in the PL spectra, the PL was significantly modified by the addition of N into the QDs. Figure 3 shows the PL spectra at 4 K for GaP samples embedded with InP QDs or InNP QDs, and the inset to Figure 3 shows a schematic band diagram for the In(N)P/GaP QD system based on a published framework.²⁰ The PL was done in 4K to increase luminescence intensity and better resolve the peak positions from InP QDs and InNP QDs. The PL intensity of InNP QD-embedded samples was significantly higher than that of InP QD-embedded samples, in contrast to the GaIn(N)As/GaAs system wherein PL intensity decreases are associated with N incorporation due to N-related defects.²¹ Compressively strained InP QDs having a size comparable to the QDs embedded in the samples described herein are expected to have Γ -like quantum-confined conduction-band ground states lying at higher energies than the X-band energy of the GaP matrix, causing electrons to occupy states at the InP/GaP interface.²⁰ Therefore, electrons tend to recombine with heavy holes inside the InP QDs in a spatially indirect way (dashed arrow, Figure 3 inset). However, localized states are formed during the incorporation of N into InNP QDs,²² and the localized states hinder the thermalization of electrons into X-derived interface states prior to recombination. Electrons trapped at the Γ -like quantum-confined ground states (E_{-}^{InNP} , solid green line, Figure 3 inset) and the N-related states lying directly below (the exponential dilute-nitride band-tail states,²³ green gradient, Figure 3 inset) can recombine with the heavy holes within the QDs in a spatially direct way²⁴ (solid red arrow, Figure 3 inset), resulting in a blue-shifted PL peak and an increase in the recombination efficiency²⁵. The GaP matrix itself has an indirect bandgap and accordingly did not exhibit any appreciable PL.

The introduction of N into the QDs also significantly affected the photoelectrochemical performance of QD-embedded p-i-n GaP devices. Electrochemical measurements were performed using a three-electrode configuration using working electrodes constructed from the p-i-n GaP and In(N)P QD-embedded p-i-n GaP devices (such that the p^{+} -GaP layer was in contact with the electrolyte solution), a Pt mesh as the counter electrode, and a Pt wire poised at the solution potential as the reference electrode. The electrodes were illuminated by an ELH-type tungsten-halogen lamp that was calibrated to provide 100 mW cm^{-2} of AM 1.5 illumination at the surface of the samples. Figure 4 shows the

current-density versus potential (J - E) data for illuminated GaP and QD-embedded GaP photoelectrodes in contact with a CH_3CN solution that contained the kinetically facile one-electron ferrocenium-ferrocene ($\text{FeCp}_2^{+/0}$) redox couple and LiClO_4 as a supporting electrolyte. The kinetically facile one-electron reversible redox couple ensured that interfacial electron-transfer processes were not rate-limiting and thus allowed measurement of the intrinsic energy-conversion properties of the photoelectrodes. All of the GaP and QD-embedded GaP electrodes produced photocurrent densities $< 400 \mu\text{A cm}^{-2}$, which is low relative to the value of J_{sc} obtained from other devices prepared using similar methods²⁶ ($\sim 1 \text{ mA cm}^{-2}$), because the electrodes were only $1 \mu\text{m}$ thick, and thus did not fully absorb the incident light. The J_{sc} of the illuminated GaP photoelectrode was $\sim 0.16 \text{ mA cm}^{-2}$, while the InP QD-embedded GaP photoelectrode exhibited $J_{\text{sc}} \sim 0.23 \text{ mA cm}^{-2}$ and the InNP QD-embedded GaP photoelectrode exhibited $J_{\text{sc}} \sim 0.29 \text{ mA cm}^{-2}$. In the InAs QD-embedded GaAs system, intermediate-band absorption from embedded QDs has been shown to produce an increase in J_{sc} but reduces the open-circuit voltage (V_{oc}).²⁷ In contrast, V_{oc} was improved for the QD-embedded GaP photoelectrodes described herein, with InNP QD-embedded photoelectrodes yielding $V_{\text{oc}} = 1.14 \pm 0.01 \text{ V}$ and InP QD-embedded photoelectrodes yielding V_{oc} of $1.11 \pm 0.01 \text{ V}$, as compared to the GaP photoelectrode which yielded $V_{\text{oc}} = 0.95 \pm 0.01 \text{ V}$. These results indicate that the embedment of the In(N)P QDs into the GaP matrix did not degrade the quality of the solid/liquid or solid-state junctions in the system.

V_{oc} is related to J_{sc} as follows:

$$V_{\text{oc}} = \frac{nk_B T}{q} \ln \left(\frac{J_{\text{sc}}}{J_0} + 1 \right),$$

where n is the ideality factor, k_B is the Boltzmann constant, T is the absolute temperature, q is the unsigned charge of an electron and J_0 is the exchange current density of the diode. The embedment of In(N)P QDs into the GaP matrix led to an increase in both J_{sc} and V_{oc} (Table 1). The increase in V_{oc} , 0.16 V and 0.19 V for InP and InNP QDs, respectively, however, exceeded the increase that can be attributed solely to the logarithmic contribution of the increased J_{sc} . When In(N)P QDs were embedded in the devices, mid-gap traps were created in the space charge region. The QDs also induce generation-recombination processes with sub-bandgap energies, as will be discussed later. These

make the In(N)P devices deviate substantially from the ideal diode where the ideality factor $n=1$ corresponds to radiation processes in the neutral charge region. Therefore we attribute the increase in V_{oc} of In(N)P devices to both increase in J_{sc} and n .

Sample	J_{sc} (mA cm ⁻²)	V_{oc} (V)
No QDs	0.16±0.01	0.95±0.01
InP QDs	0.23±0.01	1.11±0.01
InNP QDs	0.29±0.01	1.14±0.01

Table 1. Summary of short-circuit current densities (J_{sc}) and open-circuit voltages (V_{oc}) measured for GaP (no QDs) and QD-embedded GaP devices in contact with a non-aqueous solution containing the $FeCp_2^{+/0}$ redox couple and $LiClO_4$ as the supporting electrolyte.

Figure 5 shows the spectral response and external quantum yield (Φ_{ext}) measured for the devices. The InNP QD-embedded devices yielded higher Φ_{ext} values than the GaP and InP QD-embedded GaP devices. At wavelengths <549 nm, for which the photon energies exceeded the bandgap of GaP, the InP QD-embedded device yielded a slightly smaller Φ_{ext} relative to the GaP device, possibly indicating that the InP QD embedment affected the quality of the junction. However, the InNP QD-embedded device showed unambiguously higher Φ_{ext} . This behavior can be attributed to the reduced strain and better crystal quality in InNP QDs due to the smaller N atom size and N lattice hardening effect²⁸. Specifically, studies have shown that the N induced lattice hardening effect slows down lattice mismatch related relaxation and improve optical performance of dilute nitrides. Such properties can also be attributed to the decreases in J_0 in devices with InNP QDs. The Φ_{ext} values of the QD-embedded GaP samples were significantly enhanced at wavelengths > 550 nm relative to the GaP samples, and Φ_{ext} for the InNP QD-embedded GaP sample showed a significant peak near 600 nm, in good agreement

with the PL data. Because the intensity of the solar spectrum is stronger for the wavelengths in the region absorbed by the In(N)P QDs ($\sim 550\text{-}650\text{ nm}$) than for wavelengths in the region absorbed by the GaP matrix ($< 548\text{ nm}$), the QDs significantly increase the amount of sunlight absorbed by the devices. These results indicate that the embedded QDs resulted in enhanced absorption and in the increased J_{sc} value observed in the J - E data. The GaP matrix itself, which was much thicker than the QD layers, absorbed most of the photons with energies higher than the bandgap of GaP, and thus the effect of QD embedment only becomes significant when the QDs absorb photons with energies below the bandgap of GaP, which otherwise would not be absorbed by the device.

The device design also contributes to the significant improvement in performance observed for the QD-embedded GaP devices relative to GaP devices without embedded QDs. The In(N)P QDs were embedded within an undoped GaP region which is completely depleted, allowing the strong built-in electric field ($\sim 4000\text{ V cm}^{-1}$) to quickly sweep the electron-hole pairs across the junction to the contact layers. The built-in electric field also assists thermal hopping of photogenerated carriers out of the In(N)P/GaP interface and N-related localized states.

In principle, the photocurrent density of the GaP devices could be further enhanced by increasing the loading of the QDs. In practice, however, only fabrication of 8–10 layers (depending on the size of the QDs) is feasible due to the accumulation of strain. Use of a strain-balancing technique whereby a tensile strain layer is introduced between the QD layers can increase the number of QD layers that can be fabricated. Previously, the strain-balancing technique has enabled more than 100 layers stack of InAs QDs²⁹, considerably increasing the photoabsorption. Although dilute-nitride GaNP could be used for tensile strain layers in the In(N)P QD-embedded GaP devices, emission from GaNP, in the range $\sim 580\text{-}650\text{ nm}$ depending on the N composition, would prevent unambiguous quantification of the photocurrent contribution of the In(N)P QDs. Although a GaNP strain-balancing layer was not used for this demonstration, we expect that such layers would lead to improved devices. With careful control of N composition and hence strain manipulation, layers of high crystal quality In(N)P QDs with GaNP strain balancing layers can be fabricated with GSMBE. In addition, if In(N)P/GaP were to be integrated

onto the already demonstrated GaP/Si platform⁹, the materials system would be more scalable and affordable.

Conclusions

InNP QDs have much stronger PL than InP QDs due to the special bandgap alignment in these QD systems. Embedment of InP and InNP QDs, particularly InNP QDs, into GaP solar cells significantly increased the amount of light absorbed by the devices without degrading the quality of the junctions, and increases in both J_{sc} and V_{oc} were observed for the QD-embedded devices. Embedment of InNP QDs within a GaP matrix may increase the utility of GaP for photovoltaic and solar fuels generation.

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Figure 1.

Figure 2.

Figure 3.

Figure 4.

Figure 5.

Figure Captions

Figure 1. Structures of GaP p-i-n devices (left) and In(N)P QD-embedded GaP p-i-n devices (right).

Figure 2. AFM images of InP QDs (left) and In(N)P QD (right). Both images are squares with sides 1 μm in length.

Figure 3. PL spectra of InP QDs and InNP QDs at 4 K obtained using an excitation wavelength of 532 nm and a pumping energy density of 1W cm^{-2} . Inset: band diagram of the In(N)P/GaP system. The solid black lines represent the Γ -derived valence bands. The dotted lines represent the X-derived band in GaP. The solid blue line represents the Γ -derived band without N and the green line represents the lower-energy band formed in InNP QDs after N incorporation. The solid red arrow indicates the Γ - Γ transition while the dashed red arrow indicates the X- Γ transition.

Figure 4. Current-density versus voltage (J - E) behavior of GaP (black), InP QD-embedded (red) and InNP QD-embedded (blue) electrodes in contact with a CH_3CN solution containing the $\text{FeCp}_2^{+/0}$ redox couple and 1.0 M LiClO_4 as a supporting electrolyte. Illumination was provided by an ELH-type tungsten-halogen lamp calibrated to provide 100 mW cm^{-2} of simulated AM 1.5 illumination at the surface of the electrodes.

Figure 5. External quantum yield (Φ_{ext}) of GaP, InP QD-embedded GaP, and InNP QD-embedded GaP samples in contact with a CH_3CN solution containing the $\text{FeCp}_2^{+/0}$ redox couple and 1.0 M LiClO_4 as a supporting electrolyte.

Vitae

Yanjin Kuang received his Ph.D. degree in Physics at the University of California, San Diego (UCSD) in 2014 for his work on dilute nitride based intermediate band solar cell materials. Prior to his doctoral study, he obtained his M.S. degree and B.S. degree both in Physics at UCSD and at University of Science and Technology of China. His research focuses on molecular beam epitaxy and optical and electrical characterization of III-V semiconductor materials in thin film, nanowire and quantum dot structures.

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Itaru Kamiya received his BS, MS, and Ph.D. in Applied Physics from The University of Tokyo, Japan. He has been a Professor at Toyota Technological Institute, Nagoya, Japan, since 2004. Prior to taking his present position, he headed the Nanomaterials Lab at Mitsubishi Chemical Corp. after working at RIKEN, Bellcore, Imperial College, and The University of Tokyo. His research interest lies in the growth of semiconductor nanostructures using epitaxial and chemical techniques, and the optical & electronic properties of these materials particularly those related to surfaces and interfaces.

Dr. Nathan Lewis, George L. Argyros Professor of Chemistry, has been on the faculty at the California Institute of Technology since 1988 and has served as Professor since 1991. He has also served as the Principal Investigator of the Beckman Institute Molecular Materials Resource Center at Caltech since 1992, is the Principal Investigator of the Joint Center for Artificial Photosynthesis, the DOE's \$122 MM Energy Innovation Hub in Fuels from Sunlight. Dr. Lewis received his Ph.D. in Chemistry from the Massachusetts Institute of Technology. He has been an Alfred P. Sloan Fellow, a Camille and Henry Dreyfus Teacher-Scholar, and a Presidential Young Investigator.

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