Two-dimensional layered semiconductor/graphene heterostructures for solar photovoltaic applications

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Schottky barriers formed by graphene (monolayer, bilayer, and multilayer) on 2D layered semiconductor tungsten disulfide (WS$_2$) nanosheets are explored for solar energy harvesting. The characteristics of the graphene–WS$_2$ Schottky junction vary significantly with the number of graphene layers on WS$_2$, resulting in differences in solar cell performance. Compared with monolayer or stacked bilayer graphene, multilayer graphene helps in achieving improved solar cell performance due to superior electrical conductivity. The all-layered-material Schottky barrier solar cell employing WS$_2$ as a photoactive semiconductor exhibits efficient photon absorption in the visible spectral range, yielding 3.3% photoelectric conversion efficiency with multilayer graphene as the Schottky contact. Carrier transport at the graphene/WS$_2$ interface and the interfacial recombination process in the Schottky barrier solar cells are examined.

Introduction

Recently, 2D layered semiconductors have been rapidly emerging as a new class of functional materials for various applications such as nanoelectronics,$^{1-6}$ solar cells,$^{7-9}$ photodetectors$^{10-12}$ and energy storage.$^{13-16}$ Compared with traditional semiconductors such as Si,$^{17}$ GaAs,$^{18}$ CdTe,$^{19}$ and CuInSe$_2$, mechanical flexibility,$^{20}$ tunable optical bandgap,$^{22}$ and high carrier mobility$^{23}$ are among the few intriguing properties that warrant considering 2D layered semiconductors as potential candidates in solar energy harvesting. Tungsten disulfide (WS$_2$), one of the members in the 2D layered semiconductor family, has been recently explored for various nanodevice applications.$^{24-26}$ Its tunable optoelectronic properties offer unique opportunities in developing efficient solar photovoltaic technology. The major advantages of 2D layered semiconductors, such as improved photo-response and layer-dependent optical/electrical properties, have been reported for various optoelectronics applications.$^{27,28}$

It is evident that by changing the number of layers of 2D semiconductors in a stack, photons with a wide spectral range can be harvested to generate electrons. In 2D semiconductors, weak van der Waals forces bring individual atomic planes together, while strong covalent bonds exist within each crystal plane.$^{29}$ This allows for easy exfoliation of 2D crystals to make devices. However, the exfoliation process$^{30,31}$ yields sheets with only a small area. Recently, the chemical vapour deposition (CVD) method has been demonstrated for large-area growth.$^{32-34}$ On the other hand, the superior properties of graphene such as high carrier mobility$^{35}$ and negligible photo-absorption$^{36}$ in the visible spectral range are considered to be key merits in serving as a transparent conductor for solar cells.$^{37}$ The characteristics of graphene vary to a great extent among monolayer, bilayer, and multilayer graphenes.$^{38}$ Graphene stacked with 2D semiconductors in heterostructures has been explored for applications in electronics to demonstrate the possibility of incorporating the advantages of both materials.$^{39-42}$

Schottky barrier solar cells are relatively simple in the structure and fabrication process, as compared with p–n, p–i–n, and tandem solar cells, having only a metal–semiconductor junction to establish a built-in electric field in the depleted semiconductor region. Recently, extraordinary photon absorption has been reported in 2D semiconductors in only a few atomic layers.$^{43}$ In Schottky barrier solar cells, a strong electric field at the metal/semiconductor junction, a defect-free interface (to avoid Fermi level pinning), efficient photo-absorption and carrier transport are key factors to achieve high performance. In general, 2D layered semiconductors exhibit self-terminated atomic planes with no unsaturated dangling bonds at the surface, which helps in facilitating the formation of a high-quality interface and to achieve efficient carrier transport with negligible interfacial traps. In this research paper, we study the characteristics of the graphene/WS$_2$ Schottky barrier in the solar cell structure. The impacts of graphene layered configurations on the photovoltaic performance of the Schottky barrier solar cell are examined.
Experimental

CVD growth of WS₂ nanosheet

Tungsten (W) film with a thickness of ~30 nm was deposited onto SiO₂-coated Si substrates by electron-beam evaporation. The W-coated substrates were placed downstream in the CVD quartz tube chamber. Sulfur powder (99.99% purity) was placed upstream in the chamber. Sulfur was evaporated at 115 °C with 200 sccm flow of argon as the carrier gas. The chamber pressure was maintained at 1 Torr. Solid sulfur vapour resulted in WS₂ nanosheets on Si/SiO₂ substrates. A 1000 °C post-annealing treatment was performed for one hour to improve the crystallinity of WS₂ samples. Fig. 1(a) shows the schematic diagram of the CVD furnace set-up used for WS₂ growth. Note the two temperature zones used for sulfur evaporation and WS₂ growth. Fig. 1(b) is the photograph of a typical sample showing large-area growth of WS₂ on SiO₂. Fig. 1(c) illustrates the planar hexagonal lattice structure of WS₂ composed of W and S atoms in the 2D crystal plane.

Polymethyl methacrylate (PMMA) was coated on WS₂, as the supporting material to facilitate the transfer process. The sample was annealed at 90 °C for 5 minutes. Potassium hydroxide (KOH) was used to etch the SiO₂ layer, releasing WS₂ nanosheets from Si substrates and making it free-floating in KOH. The free-floating WS₂ nanosheet was manually transferred from KOH to aluminum (Al) coated glass substrate for solar cell fabrication. This method is feasible for large-area grown samples of 2D semiconductors. The etching rate of SiO₂ is determined by the KOH concentration. A fast etching rate may affect the material quality of 2D semiconductors. Therefore, it is important to use a seamless etching process to minimize process-induced effects on WS₂ nanosheets.

CVD growth of graphene

Growth of monolayer and multilayer graphene was performed by the CVD method using Cu and Ni as the substrates, respectively. Methane and hydrogen gases were used to take advantage of the solubility of carbon in Cu and Ni substrates, resulting in monolayer and multilayers of graphene, respectively. The growth temperature was fixed at 1000 °C for both monolayer and multilayer graphene growth. The PMMA layer was coated on the graphene-grown Cu and Ni substrates by the spin coating method at 2000 rpm for 1 minute, followed by a baking at 90 °C for 5 minutes. The Cu and Ni substrates were etched by a FeCl₃ solution. The separated graphene films were then transferred to deionized water to remove excessive FeCl₃ on the surface. Subsequently, the PMMA-supported monolayer and multilayer graphene layers were transferred onto WS₂ nanosheets for solar cell fabrication. After transferring graphene, the PMMA was dissolved by acetone.

Material characterization

Raman spectroscopy with a laser excitation wavelength of 532 nm (Horiba Scientific) was performed to study the as-grown WS₂ nanosheet, and monolayer/bilayer/multilayer graphene. The WS₂ nanosheet, coated on a glass substrate, was subjected to UV/visible absorption measurement to acquire the values of optical bandgap and transmittance using a Camspec M550 Double-Beam Scanning UV/Vis Spectrophotometer (in the spectral range of 200 nm–900 nm). The surface profile of the WS₂ nanosheet was analysed by Dimension 3100 atomic force microscopy (AFM).

Solar cell fabrication and characterization

Al thin films of 100 nm were deposited on three glass substrates by the electron beam evaporation method. The CVD-grown WS₂ nanosheets were transferred onto each of the Al-coated glass substrates by a manual transfer process. Monolayer, bilayer and multilayer graphene sheets, synthesized by the CVD method, were subsequently transferred onto the three Al-coated substrates. The stacked structures glass/Al/WS₂/graphene were annealed at 100 °C for five minutes to remove any excess solvents present in the samples. In our experiment, monolayer and multilayer graphene sheets were as-grown on Cu and Ni foils, respectively. The bilayer graphene was obtained by stacking two monolayer graphene grown on Cu. Three Schottky barrier solar cells, having different stacked structures of (i) glass/Al/WS₂/graphene, (ii) glass/Al/WS₂/bilayer graphene, and (iii) glass/Al/WS₂/multilayer graphene, were characterized to explore the material interfacial properties and photovoltaic characteristics. Fig. 2 shows a combined schematic view of the fabricated three Schottky barrier solar cells employing WS₂ nanosheet as a photoactive semiconductor with monolayer, bilayer, and multilayer graphene as the Schottky contacts. The current-density-vs.-voltage (J-V) characteristics of the Schottky barrier solar cells were measured under both dark conditions and standard AM 1.5 illumination using an Agilent B1500A semiconductor device analyzer. A calibrated Xe arc lamp was used to simulate the natural solar light. The Schottky barrier solar cells were illuminated through the graphene contacts as shown in Fig. 2.
Results and discussion

AFM characterization was performed on the CVD-synthesized WS$_2$ nanosheet to examine the surface properties. A 30 × 30 μm$^2$ surface topography scan on the WS$_2$ nanosheet grown on Si/SiO$_2$ is shown in Fig. 3(a). A clear difference is observed between the SiO$_2$ surface and the CVD WS$_2$ nanosheet. In the same scanning area, the phase image was generated, showing significant contrast between the two materials, as displayed in Fig. 3(b). The AFM line scan measurement was also conducted on the WS$_2$ nanosheet sample. The average value was considered on the SiO$_2$ substrate and on the WS$_2$ nanosheet, confirming the thickness of the WS$_2$ nanosheet at about ∼37 nm, as shown in Fig. 3(c).

Raman spectroscopic measurements were performed to study the monolayer, bilayer and multilayer graphene sheets. The major characteristic peaks of monolayer graphene were observed at 1582 cm$^{-1}$ and 2679 cm$^{-1}$. The intensity ratio between G and 2D peaks confirms it is monolayer graphene, as shown in Fig. 4(a). In the case of bilayer graphene, the intensity of G and 2D peaks is observed to be nearly the same, as seen in Fig. 4(b). Multilayer graphene sheets obtained on Ni foil showed dominant G peaks, as illustrated in Fig. 4(c). The intensity ratio between 2D and G peaks of monolayer, bilayer, and multilayer graphene sheets was calculated and shown in Fig. 4(d). It is well known that the 2D peak in graphene originates from double resonance electron-phonon scattering which exhibits significant differences among monolayer, bilayer, and multilayer graphene sheets.

It is expected that the intensity of the G peak increases with the number of graphene layers. As shown in Fig. 4(a)–4(c), the G peak of multilayer graphene is the strongest among all graphene samples. The position and full-width-at-half-maximum of the G-peak are not sensitive to the number of layers. It originates from in-plane vibration of sp$^2$ carbon with doubly degenerated phonon modes at the center of the Brillouin zone. However, the 2D peak shows significant variation due to the change in the band structure of multilayer graphene. We also observe that all three samples do not show a noticeable D peak in the Raman spectra, confirming a reasonably good quality of graphene. It should be noted that the ratio of 2D and G peaks decreases with the number of graphene layers, as seen in Fig. 4(d). The ratio is much higher for monolayer graphene as compared with bilayer and multilayer graphene sheets, with...
multilayer exhibiting the smallest value among all samples, as illustrated in Fig. 4(d).

Two strong active vibrational modes of WS$_2$ were observed. We confirm the major vibrational Raman modes in the CVD as-synthesized samples from the disappearance of the peak at $\sim$421.3 cm$^{-1}$ ($A_{1g}$ mode). The peak at $\sim$352.8 cm$^{-1}$ with no polarization dependence is the $E_{2g}$ mode. Fig. 5(a) shows the Raman spectrum obtained on the WS$_2$ nanosheet samples in the wavenumber range of 300 cm$^{-1}$–500 cm$^{-1}$. The $E_{2g}$ peak (352.8 cm$^{-1}$) represents in-plane atomic vibration in the WS$_2$ lattice, while the $A_{1g}$ peak (421.3 cm$^{-1}$) is associated with the out-of-plane vibrational mode. The $E_{2g}$ mode involves in-plane displacement of both transition metal and chalcogen atoms, while the $A_{1g}$ mode involves only chalcogen atoms. A schematic of the two Raman active vibrational modes of WS$_2$ is illustrated in Fig. 5(b). The values obtained for the representative peaks are slightly different, when compared to those of the bulk crystals (355 cm$^{-1}$ and 421 cm$^{-1}$, respectively). The very slight shift might be from the unintentional impurity contents present on the surface.

The Lorentzian double-peak-fit was performed on the Raman spectra obtained from the three nanosheet samples to extract the values of FWHM, and a typical fitting is shown in Fig. 5(c), performed on sample 2. The two Lorentzian fits yield the peaks corresponding to the experimental data at both signature values. All the three samples showed FWHM values approximately $\sim$18 for $E_{2g}$ and 8 for $A_{1g}$, respectively, suggesting that the CVD-grown WS$_2$ nanosheets are highly crystalline in nature with negligible structural deformation (that could be introduced during the SiO$_2$ wet etching step and the subsequent transferring process between different substrates). The post-treatment, i.e., annealing at 1000 °C after growth, helps in forming the highly crystalline nanosheets. Fig. 5(d) shows the calculated difference in the wavenumber of $E_{2g}$ and $A_{1g}$ vibrational modes for the three samples. The nearly-equal difference values in wavenumber suggest that the three nanosheet samples possess nearly identical characteristics, synthesized by the same growth process.

Three WS$_2$ nanosheets were consecutively transferred onto the glass substrate for conducting characterization using UV-visible optical absorption spectroscopy. From the AFM profiling measurement, it is confirmed that each nanosheet is about $\sim$37 nm thick and hence the transfer process yields a 111 nm stack. The WS$_2$ stack for UV visible absorption measurement was used for subsequent fabrication of a solar cell with graphene as a Schottky contact. Fig. 6 shows the measured optical transmittance of the WS$_2$ nanosheet in the wavelength range of 200 nm–900 nm. The nanosheet exhibits a maximum optical transmittance of $\sim$43% in the longer wavelength region, while showing very efficient photo-absorption in the visible spectral window. The optical absorbance spectrum corresponding to the transmittance is shown in Fig. 6. A strong photo-absorption in the wavelength range starting from 200 nm to 900 nm suggests that the WS$_2$ nanosheet is desirable for solar energy harvesting. We also calculated the optical bandgap of the WS$_2$ nanosheet from the measured optical transmittance using the Tauc plot method via the relationship $\alpha h \nu \propto (h \nu - E_G)^{1/2}$, where $\alpha$ is the absorption coefficient of the WS$_2$ nanosheet, $h \nu$ is the incident photon energy, and $E_G$ is the optical bandgap (that determines the photo-absorption spectral range). The bandgap is calculated to be $\sim$1.3 eV from the transmittance spectrum shown in Fig. 6. A dominant photo-absorption associated with a 1.3 eV bandgap is highly favourable for solar cell design, as this spectral range suits well thin-

Fig. 5 (a) Raman spectra of three CVD synthesized WS$_2$ nanosheet samples used for solar cell fabrication showing two major vibrational modes. (b) Schematic of atomistic view of the two modes illustrating the direction of atomic movements. (c) Double peak Lorentzian fit of the Raman spectrum for sample 2. (d) The calculated wavenumber difference of two vibrational modes for three WS$_2$ nanosheet samples.

Fig. 6 Measured UV-visible absorbance and transmittance of the WS$_2$ nanosheet exhibits strong photo-absorption in the spectral window of 200 nm–900 nm and the corresponding transmittance shows a maximum value of $\sim$43% at a longer wavelength region with a calculated optical bandgap value of 1.3 eV.
film semiconductors such as amorphous Si, CdSe, CdTe and CdS.

Three Schottky barrier solar cells (cell-1 with monolayer graphene on WS$_2$, cell-2 with bilayer graphene on WS$_2$, and cell-3 with multilayer graphene on WS$_2$) were characterized under dark conditions and AM1.5 illumination. The graphene induced Schottky barrier in the WS$_2$ nanosheet is used to develop an electric field at the graphene/WS$_2$ junction, used for efficient electron–hole pair dissociation in the solar cells. The built-in electric field at the graphene–WS$_2$ junction points towards the graphene contact from the WS$_2$. Thus, the electron transport occurs in the opposite direction, from the WS$_2$ to the Al ohmic contact. The holes generated in the valence band of the WS$_2$ nanosheet are transported towards the graphene contact along the electric field direction.

Fig. 7(a) shows the $J$–$V$ characteristics of the three devices under AM1.5 illumination calibrated by a standard solar cell. The output power characteristics as a function of the applied voltage are shown in Fig. 7(b). The Schottky-barrier solar cell employed monolayer graphene as a contact on WS$_2$ (cell-1) exhibits the key device metrics of $J_{SC}$ (short-circuit current density), $V_{OC}$ (open-circuit voltage), $P_{MAX}$ (maximum power), and $\eta$ (photo-conversion efficiency) of 9.2 mA cm$^{-2}$, 640 mV, $1.6 \times 10^{-3}$ W, and 2.3%, respectively. The cell-2 (bilayer graphene contact on WS$_2$) shows $J_{SC}$, $V_{OC}$, $P_{MAX}$ and $\eta$ as 14.2 mA cm$^{-2}$, 680 mV, $2 \times 10^{-3}$ W, and 2.9% respectively. For cell-3 (multilayer graphene contact) the $J_{SC}$, $V_{OC}$, $P_{MAX}$ and $\eta$ are 16 mA cm$^{-2}$, 700 mV, $2.3 \times 10^{-3}$ W, and 3.3% respectively. It should be noted that the changes in solar cell performance metrics are influenced only by the graphene contact (made of varying number of layers), as the WS$_2$ nanosheet thickness remains the same. While improvement of solar cell performance is expected by increasing the WS$_2$ nanosheet thickness, the presented effort is focusing only on the effect of graphene contacts.

The $J_{SC}$ characteristics of the Schottky barrier solar cells were measured under different illumination conditions, as shown in Fig. 7(c). A nearly linear increase in $J_{SC}$ with incident light intensity is observed for all three devices. The photo-carrier generation is significantly impacted by the incident light intensity, therefore drastically changing the $J_{SC}$ value. A larger increase in $J_{SC}$ is observed in the Schottky barrier solar cell with multilayer graphene contact as compared with other two devices, suggesting that multilayer graphene would be the desirable contact configuration for the Schottky barrier solar cell.

Fig. 7(d) summarizes the measured key metrics of the solar cell ($J_{SC}$, $V_{OC}$ and $\eta$) for the three Schottky barrier solar cells with monolayer, bilayer, and multilayer graphene contacts. Cell-3 shows exceptional value in $V_{OC}$ as compared with cell-1 and cell-2, offering an insight about the major change in the quasi-Fermi level splitting due to optical illumination. The difference in the $J_{SC}$ value among all three solar cells would be due to efficient electron–hole pair separation and collection as a result of significant modification in Schottky barrier behaviour by bilayer and multilayer graphene contacts. Therefore, we attribute the distinction in overall photovoltaic performance in the solar cell to the critical impact of graphene contacts and the associated Schottky junction properties.

Fig. 8 shows the energy-band diagram of the graphene–WS$_2$ nanosheet Schottky barrier solar cell (as shown in Fig. 2) with reference to the vacuum level ($E_0$). Photons incident on the device through the graphene layer and excite electrons from the valence band ($E_v$) to the conduction band ($E_C$) of the WS$_2$ nanosheet, leaving holes in the valence band. The electric field ($E_J$) developed at the graphene–WS$_2$ interface, pointing towards the graphene contact, drives the photo-generated electrons towards the Al electrode. The parameters $\phi_G$, $\phi_{WS_2}$ and $\phi_B$ represent the work function of graphene, WS$_2$, and the Schottky

![Fig. 7](image1.png)  
**Fig. 7** Schottky barrier solar cell performance: (a) $J$–$V$ characteristics of three solar cells under AM1.5 illumination, (b) output power, (c) light intensity dependence of $J_{SC}$ values, and (d) comparison of performance metrics showing $J_{SC}$, $V_{OC}$ and $\eta$ for different solar cells.

![Fig. 8](image2.png)  
**Fig. 8** Schematic of energy level alignment at the graphene–WS$_2$ interface illustrating Schottky barrier height, junction electric which drives electron–hole transport towards their respective electrodes.
barrier height, respectively. The parameter $E_F$ represents the position of the Fermi level for graphene, the WS$_2$ nanosheet and Al contact. The transport pathway of photo-generated electrons starting from $E_C$ of the WS$_2$ nanosheet to $E_F$ of the Al contact and to graphene through an external load ($R_L$) is illustrated.

In conventional bulk or thin-film semiconductors such as Si and GaAs, the surface is not self-terminated. Since the lattice periodicity is interrupted, the surface possesses a huge density of unsaturated dangling bonds which act as recombination centers limiting solar cell performance. The surface recombination velocity of photo-generated carriers is greatly influenced by the surface traps, resulting in a key recombination process. 2D layered semiconductors, by nature, are individual atomic planes coupled via weak van der Waals forces. The surface is a self-terminated atomic plane with no dangling bonds. Therefore, in principle the surface recombination velocity is expected to be extremely low, as compared to conventional bulk and thin-film semiconductors. The low surface recombination velocity helps in facilitating an efficient photo-generated carrier collection at the electrode, yielding solar cells with significantly improved photocurrent.

The $J$–$V$ characteristics under dark conditions represent the interface material quality that directly impacts carrier transport and recombination in solar cells. Electron transport under dark conditions is in the opposite direction to that under illumination. If dark current is minimized, the net photocurrent increases. Fig. 9(a) shows the dark $J$–$V$ characteristics of the three Schottky barrier solar cells in a forward-bias region of 0 V–1 V. The onset voltage of the dark current is lower in cell-3 (multilayer graphene contact), and is shifted by about a hundred meV in the positive bias region for cell-2 and cell-3, as shown in Fig. 9(a). The reduced dark current density, positive shift of the onset voltage, and increased photocurrent indicate that the multilayer graphene contact helps in reducing electron capturing from the conduction band of the WS$_2$ and hole capturing from the valence band, and eventually controls the trap-assisted recombination process in the WS$_2$. Fig. 9(a) inset shows the semi-log plot of the dark $J$–$V$ characteristics of the three solar cells, showing distinct behavior in the forward bias region (from 0 V–1 V).

The dark current density is analyzed in the low- (0.2 V–0.4 V) and medium-forward-bias regions (0.5 V–0.8 V) to study the recombination possibilities at the bulk of WS$_2$ and graphene–WS$_2$ interface, as well as the impact of graphene on carrier transport and recombination. Fig. 9(b) and (c) show the curve fitting by a single-diode model performed on all the three devices in the low- and medium-forward-bias regions, respectively. The linear region in the semi-log plot of $J$–$V$ characteristics is fit into the single-diode model to extract the dark saturation current density ($J_0$),

$$J_D(V) = J_0 \left( e^{\frac{V}{kTq}} - 1 \right)$$

Here $q$ is electronic charge, $V$ is the applied voltage to solar cell, $n$ is the ideality factor, $k$ is the Boltzmann constant, and $T$ is temperature at which the solar cell operates (300 K).

The values of $J_0$ for three solar cells with different graphene contacts are listed in Table 1. In both conditions, low- and medium-forward-bias regions, cell-3 with multilayer graphene contact exhibits lower $J_0$ than other two devices. In particular, the diode model fitting at the medium-forward-bias region showed a significantly lower $J_0$ value for cell-3 than that of cell-1 and cell-2, demonstrating a better control over dark current in device configuration with multilayer graphene contact. This could be the primary cause for the improved photovoltaic performance of the cell-3 as compared with other two devices.

It is known that electron transport occurs in opposite directions under dark and illumination conditions. Lower dark current value confirms the improvement in the illuminated $J$–$V$ characteristics of a solar cell. In other words, the net photocurrent of a solar cell under illumination conditions would be

<table>
<thead>
<tr>
<th>Devices</th>
<th>$J_0$ (A cm$^{-2}$) Low forward bias</th>
<th>$J_0$ (A cm$^{-2}$) Medium forward bias</th>
</tr>
</thead>
<tbody>
<tr>
<td>Cell-1</td>
<td>$7.7 \times 10^{-7}$</td>
<td>$7.3 \times 10^{-7}$</td>
</tr>
<tr>
<td>Cell-2</td>
<td>$6.2 \times 10^{-7}$</td>
<td>$2.9 \times 10^{-8}$</td>
</tr>
<tr>
<td>Cell-3</td>
<td>$5.0 \times 10^{-7}$</td>
<td>$9.3 \times 10^{-10}$</td>
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adversely affected by an increase in dark current. This can be explained, in a simple way, by eqn (2),

\[ J(V) = J_{SC} - J_D(V) \]  

(2)

Here, \( J_0 \) is the dark-current density of the solar cell and \( J_D(V) \) is the dark-current density as expressed in eqn (1). It is evident that decreasing \( J_0 \) would increase the net photocurrent, therefore improving the photovoltaic performance. The Schottky barrier solar cell with multilayer graphene on WS\(_2\) exhibits the lowest dark saturation current density, as listed in Table 1, leading to enhanced photoconversion efficiency. The \( V_{OC} \) increases of 20 mV and 60 mV are observed in cell-3 with multilayer graphene contact as compared with cell-2 (bilayer graphene contact) and cell-1 (monolayer graphene contact), respectively, which can be attributed to the better control over dark current using multilayer graphene contact on the WS\(_2\) nanosheet. This can be explained using the following eqn (3):

\[ V_{OC} = \frac{kT}{q} \ln \left( \frac{J_{SC}}{J_0} + 1 \right) \]  

(3)

It is therefore understood that better capability to suppress dark current can be achieved via using highly conductive multilayer graphene, as compared with monolayer and bilayer graphene, which results in improved solar cell performance. The advantage of adopting multilayer graphene as the Schottky contact on the layered semiconductor WS\(_2\) nanosheet is demonstrated through both illuminated and dark \( J-V \) characteristics and the relationship that connects both in eqn (1)–(3).

Conclusions

We demonstrate graphene/2D semiconductor Schottky barrier solar cells with different layer configurations of graphene contacts on the CVD-assembled WS\(_2\) nanosheets. The solar cell performance is significantly improved by incorporating multilayer graphene as the Schottky contact. While the WS\(_2\) nanosheet exhibits the same efficient visible spectral photo-absorption in all three demonstrated solar cells, the multilayer graphene helps in suppressing the dark current much better than monolayer and bilayer graphene, which improves the Schottky barrier solar cell performance. The experimental results point to the possibility of designing all-2D-layered-material-based solar cells with potential advantages in efficient photovoltaic conversion, mechanical flexibility, and low material consumption.

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