Nanoscale electrical characterization of graphene-based materials by atomic force microscopy

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Graphene, an atomically thin two-dimensional (2D) material, exhibits outstanding electrical properties and thus has been employed in various electronic devices. However, the device performance strongly depends on the structural variations present in the graphitic lattice, such as crystal domains, grain boundaries, lattice imperfections, dopants, etc., which are nanoscopic in nature. Hence, understanding the correlation between the structure and the electrical properties in the nanoscale is essential.

Atomic force microscopy (AFM) techniques provide the best way to picture such relationships, which is particularly in demand for future miniaturized devices. This review article highlights the characterization of the electrical properties of graphene-based materials via AFM-based techniques such as conductive AFM, scanning Kelvin probe microscopy, electrostatic force microscopy, and piezoresponse force microscopy that is certainly beneficial for a broad research community not only working on graphene-based materials but also in the fields of other 2D materials and scanning probe microscopy.

Introduction

Graphene is one of the most studied materials during the past decade spanning its branches in almost all research areas [1, 2]. Despite its atomically thin lattice structure, defect-free and highly ordered graphene sheets exhibit unique properties, such as high electrical and thermal conductivity, good optical transparency, and excellent mechanical strength useful in nanoelectronics applicable in diversified fields [3–5]. Recently, researchers have found that twisted bilayer and trilayer graphene show exotic superconductivity offering new insights into the field of graphene-based electronics [6, 7]. Pristine graphene isolated by the mechanical exfoliation (ME) of graphite and graphene grown by chemical vapor deposition (CVD) and epitaxial method possess the best electrical and mechanical properties due to the intact lattice structure [8–10]. On the other hand, the electrical conductivity of graphene synthesized by the reduction of graphene oxide – GO (known as reduced graphene oxide – RGO) is very much lower than graphene fabricated by above mentioned methods due to the structural imperfections caused by chemical and/or thermal treatments exploited during their synthesis. Yet, it is widely studied as it is more economical and its chemical and physical properties can be tuned depending on its end use [11–15]. In this review, nanoscale characterization of the electrical properties, such as conductivity, surface potential, work function, and piezoresponse, of graphene synthesized by all these methods will be discussed.

Depending on the fabrication method, the morphology of the substrate, and growth conditions, nanoscale inhomogeneities may present in the graphene lattice, affecting the macroscopic electronic behavior of graphene-based devices [16, 17]. Even for graphene fabricated in the same manner, nanoscale structural variations, such as grain boundaries, crystal domains, wrinkles, multilayers, folded edges, cracks, adsorbed molecules, and dopants, are expected [18, 19]. Moreover, the zero-bandgap...
nature and extremely high conductivity in defect-free graphene make it hard to be used in graphene-based transistors [20]. Hence, nanostructuring of the graphene lattice is necessary to tailor the electrical properties depending on the applications. RGO is more complicated as it has plenty of lattice defects that vary depending on the initial GO structure and reduction method [21]. Oxygen-containing groups, lattice vacancies, and stone-wales defects are a few to mention. Since monolayer or few-layer graphene are used in nanoelectronic devices, such nanoscale structural defects or modifications highly affect the device performances. Scanning probe microscopy (SPM) techniques offer a way to characterize such structural variations and changes in electrical and mechanical properties simultaneously at nanoscales [22, 23].

SPM is an advanced imaging technique down to atomic scale that can characterize almost all types of materials, not only their morphology but also their electrical or mechanical properties [23]. It utilizes a sharp tip that scans over the sample and senses the changes in the sample topography. Two broad categories of SPM techniques are available; scanning tunneling microscopy (STM), which detects the change in the tunneling current between the tip and the sample, and atomic force microscopy (AFM), which detects the change in the local forces between the tip and the sample (see Fig. 1) [24]. Compared to STM-based electrical characterization techniques, AFM-based techniques provide more choices for measuring electrical properties, namely, conductivity, current distribution, surface potential, and piezoelectric response along with the morphology of the sample under investigation [25–27]. Moreover, AFM measurements can be done at any atmospheric condition and are not limited to conductive samples. This allows the researchers to study the nanoscale characteristics of even insulators, semiconductors, and biological samples with high special resolution [28, 29]. When comes to 2D materials such as graphene, transition metal dichalcogenides (TMDCs), and hexagonal boron nitride (h-BN), characterization of electrical properties in nanoscale is very important to understand the physical/chemical interactions between layers and structural modifications [30, 31]. There are several review papers on different SPM techniques and their principle, usage in 2D materials, and organic devices [32–34]. All of these are written for a broader community and some necessary aspects are not sufficient. Therefore, a more focused article reviewing the applicability and advantages of nanoscale electrical characterization of various graphene-based materials will be helpful for the readers, not only who are working on graphene-based materials but also on other 2D materials. Furthermore, graphene itself is a family with materials having diverse structural and electrical properties, which should be carefully treated prior to and during the measurements. Thus, this review article gives an overview of AFM-based characterization techniques such as conductive AFM (C-AFM), scanning Kelvin probe microscopy (SKPM), electric force microscopy (EFM), and piezoresponse force microscopy (PFM) that have been used in characterizing electrical properties in graphene-based materials and the future perspectives. AFM-based techniques have also been advanced by combining with other characterization techniques such as tip-enhanced Raman spectroscopy (TERS) and AFM-IR (infra-red) spectroscopy so that the nanoscale structural properties can be visualized as a map [35, 36]. However, since the main scope of this review article is the electrical property characterization of graphene via AFM techniques, the readers are encouraged to refer to other review articles for detailed explanations of the working principle of the techniques [24, 34, 37]. Another advanced technique to mention is AFM lithography, which is mainly employed in nanoscale patterning of samples in order to modify the structural and electrical properties [38]. This will be discussed in this review along with electrical property measurements.

![Basic setup of STM and AFM](https://example.com/basic_setup.png)

**Figure 1**: Basic setup of STM and AFM. STM detects the change in the tunneling current between the tip and the sample, whereas AFM detects the change in the local forces between the tip and the sample.
Investigation of current distribution by conductive atomic force microscopy (C-AFM)

Introduction to C-AFM
Generally, the sheet resistance or conductivity of thin films is measured by more direct methods such as the four-point probe method, which is quick and easy [39]. Since these methods give the spatially averaged electrical measurements of the film, it is necessary to find a way to characterize local electrical characteristics of the material at a nanoscale. In this regard, C-AFM offers a way for simultaneous imaging of nanostructures and electrical properties and it has been employed in graphene-based electronic devices, organic semiconductors, and photovoltaic heterojunctions, self-assembled monolayers, etc. [40]. In C-AFM, a sharp, metal-coated conductive probe is placed in contact with the sample, a bias voltage is then applied between the probe (movable electrode) and the counter electrode (in contact with the sample), and the resulting current is measured [41, 42]. The standard C-AFM operation mode is contact mode, where the tip and the sample are in permanent physical contact. Besides this, dynamic operation modes such as PeakForce TUNA (from Bruker) and torsional resonance mode, where the tip oscillates closer to the sample surface, have been developed to expand the usability of C-AFM for characterizing novel materials and applications [43, 44]. Yet, the standard contact mode C-AFM is prominent over other modes. As shown in Fig. 2, two standard configurations of C-AFM are in use [40]. In the vertical geometry, the sample is deposited on a conductive substrate and the out-of-plane current is measured. Whereas in horizontal geometry, the sample is deposited on an insulating substrate and the in-plane current is measured. Through C-AFM, the current either can be obtained as a map by scanning over the sample under a constant voltage or can be recorded as a function of voltage to obtain I-V characteristics. Hence, one can have a better understanding of the current distribution of the sample with respect to the nanoscale structural variations. Moreover, in C-AFM, force feedback is used to control the probe position (in contrast to the current feedback in STM) enabling to study electrically insulating regions [24].

During C-AFM, there are a few things to be taken into consideration. Since C-AFM is carried out in contact mode, the selection of a suitable cantilever, depending on the sample morphology or hardness, is important. Tip sharpness, tip–sample contact area, contact resistance, and wear resistance are sensitive parameters that determine the absolute current value [45]. Moreover, contaminations on the tip apex and surrounding environmental conditions, such as humidity, temperature, and external vibrations, are also important parameters. On the other hand, the underlying substrate (on which the sample is placed) is also believed to influence the local conductivity of the sample [16, 46]. The physical contact area ($A_c$) between a tip and the sample is generally estimated by the Hertz contact theory, which represents as,

$$ A_c = \pi r_c^2 = \pi \left( \frac{F_c R_{tip}^3}{4 E_{tip} \left(1 - \nu_{tip}^2\right)} + \frac{1 - \nu_{sample}^2}{E_{sample}} \right)^{\frac{3}{2}} $$

Here, the tip contact radius and tip radius are $r_c$ and $R_{tip}$ respectively. The $F_c$ is the force between the tip and the sample, which is the sum of the applied force and the adhesion force. The $E$ and $\nu$ are the elastic modulus and the Poisson’s ratio of tip and sample, respectively [47]. Accordingly, tip radius and the contact force play a huge role in C-AFM yet, determining a real value for $F_c$ would be challenging as it depends on additional forces as well. Figure 3(a) shows the schematic of the Hertzian contact model, where a spherical AFP tip indenting into a flat surface. Generally, the actual radius of the AFM tip can be judged from its scanning electron microscope (SEM) image (even though the average radius is given by the manufacturer). Applied force and the adhesion force can be determined during measurement and from the force–distance curve, respectively. Elastic modulus and the Poisson’s ratio are material properties. Substituting these values into Hertz equation, $A_c$ can be estimated. Depending on the C-AFM geometry, different resistances can affect to the total resistance or conductivity of the sample. According to a model

![Horizontal geometry](image1.png)

![Vertical geometry](image2.png)

**Figure 2**: C-AFM measurement geometries. Horizontal geometry, where the sample is placed on an insulating substrate and the in-plane current is measured. Vertical geometry, where the sample is placed on a conductive substrate and the out-of-plane current is measured.
reported by Giannazzo et al. given in Fig. 3(b), four types of resistances contribute to the total resistance ($R$) of the sample: tip contact resistance on to graphene ($R_{c,\text{tip}}$), spreading resistance related to the local resistance of graphene, which depends on the tip radius ($R_{\text{tip}}$), series resistance related to average graphene sheet resistance ($R_{\text{series}}$), and macroscopic contact resistance of the Ohmic contact, made by Ni/Au bilayers ($R_{c,\text{macro}}$) [42]. The total contact resistance originated from the contacting interfaces of electrical contacts, such as tip–sample and electrode-sample interfaces, can be determined by measuring the resistance of the film as a function of tip-electrode separation [48, 49]. The intercept of the fitted curve of resistance versus tip-electrode separation, gives the contact resistance of the total system. Nonlinearities in the plot can be originated due to the irregularities present in the sample.

**Characterization of graphene-based materials by C-AFM**

The C-AFM plays a remarkable role in depicting the electrical and mechanical properties of graphene-based materials. It is useful not only to characterize nanoscale electronic properties but also to investigate atomic-scale friction properties, reduction of graphene oxide (GO), oxidation and hydrogenation of pristine graphene, etc. Depending on the fabrication method and the type of substrate used, the resulting graphene shows different structural and electrical properties. Therefore, C-AFM is highly suitable for nanoscale electrical characterization of graphene [50]. Also, C-AFM has also been employed in probing the current transport in graphene-based van der Waals heterostructures [51]. However, in this section, we will give the attention to the local electrical characterization of graphene, graphene oxide (GO), and reduced graphene oxide (RGO). In an extensive study done on C-AFM-assisted electrical property measurement of highly oriented pyrolytic graphite (HOPG), Banerjee et al. presented that a few interesting phenomena can be identified when the graphene layers are strained and dislocated from the bulk [18, 52]. The top graphene layers always showed higher current and graphene edges showed either sharp increments or sharp dips in current. The graphene sheets are inhomogeneously exfoliated during mechanically exfoliated from the bulk HOPG exerting vertical and lateral stresses and hence, the sheets will get dislocated having different c-axis distances. As a result, the $\pi$-electrons of the loosely bound top layers will have high mobility leading to higher conductivity. Additionally, the sharp increase or decrease in current at the edges is attributed to zig-zag and armchair edges, respectively. In two different reports, Ahmed et al. presented a comparative study on how the local conductivity of ME-graphene and CVD-graphene changes at the nanoscale [46, 53]. It was found that there is a drastic drop in current at domain boundaries, step edges, and wrinkles. Also, the conductivity of ME-graphene is higher compared to CVD-graphene. Figure 4(a)–(c) shows the topography, current, and the lateral force microscopy (LFM) images of ME-graphene deposited on SiO$_2$/Si substrate, respectively. The current and LFM images show better contrast compared to the topography image. The bright parts are the Au electrodes and the graphene sheet with edges (green and red lines) and step edges (black dotted line) located on either side of the electrode. The red-colored arrow in the LFM image indicates a crack on the graphene film that has disconnected the film from the electrodes resulting in no current map in that area (area under the black-colored circle in the current image). This indicates that
the continuity of the graphene flake is important to detect the current map as it is deposited on an insulating substrate. The dark area in the current image is attributed to the insulating SiO$_2$/Si substrate and the horizontal lines are attributed to the unstable electric contact between the sample and the tip (sample bias 300 mV). The current profiles were taken along the lines denoted in Fig. 4(b) assigned to the difference in the edge state (not shown in here). Accordingly, the red dashed line represents a zig-zag edge as it shows a high current peak in the profile. Conversely, the edge indicated by the black dotted line shows that the bottom layer has a lower current than the top layer, which has been described as a result of the interlayer interaction. They have also suggested that the presence of wrinkles decreases the electrical conductivity on graphene sheets because the electrons at the wrinkles are expected to be scattered as explained by the changed band structure.

In contrast to ME-graphene, CVD-graphene consists of monocrystalline grains joined by grain boundaries and a network of wrinkles. It has been found that the grain boundaries are more resistive compared to the conductivity at the wrinkles [53, 54]. Other than that, the graphene film is electrically homogeneous. As given in the topographic image of CVD-graphene transferred onto SiO$_2$/Si substrate [see Fig. 4(d)], wrinkles appear brighter than the flat region due to their height. On the other hand, in the corresponding current image [Fig. 4(e)], they appear darker indicating the lower current (more than 50% drop in current compared to that on the flat graphene area). The current drop along ripples is not more than 10%. The I-V curves taken next to a wrinkle and on the wrinkle (denoted as 1 and 2, respectively, in the current image) are shown in Fig. 4(f), where the current measured on the wrinkle is lower. The point contact between the AFM tip

Figure 4: (a) Topography, (b) current (at 0.1 V bias voltage), and (c) LFM images of ME-graphene. (Reproduced with permission from Ref. [46]). (d) Topography and (e) current (at 0.3 V bias voltage) images of CVD-graphene on SiO$_2$/Si substrate. W$_1$ and W$_2$ are wrinkles, and the black arrows show the ripples. (f) I–V curves taken away (1) and on (2) the wrinkle denoted in (e) and the inset shows the I–V curves in the ±0.1 V range. (g) Distribution of contact resistance measured on wrinkle ($R_{c,wr}$) and away from wrinkle ($R_{c,gr}$) (Reproduced with permission from Ref. [54]).
and graphene can be regarded as an Ohmic contact because in the voltage range ± 0.1 V both I-V curves are approximately linear. Figure 4(g) show the distribution of contact resistance measured on the wrinkle (R_{c,wr}) and away from the wrinkle (R_{c,gr}) and the values are 57 MΩ and 13.6 MΩ, respectively. Additionally, in the presence of multilayers and grains that are completely isolated or weakly connected, the current flow on the film will not be uniform. Multilayers show higher conductivity than single layers due to the effect of the underlying substrate on a single layer [53]. Hence, uniform graphene film with large domains is essential in fabricating highly conductive graphene-based films.

By incorporating a homemade current amplification and conversion device, called ResiScope to the AFM, Hauquier et al. have observed a clear distinction in resistance (R) between different number of stacked layers of ME-graphene deposited on polyaminophenylene (PAP) film grafted on Au substrate [55]. Although the graphene layers cannot be distinguished from the

![Figure 5](image_url)

*Figure 5*: (a) Topography, (b) electrical resistance, and (c) friction images of few-layer CVD-graphene sheets deposited on PAP/Au. Corresponding (d) height, (e) resistance, and (f) friction profiles along the lines in [(a)–(c)]. (g) I-V curves measured in region I (very thin graphene sheets on PAP/Au) at a very low applied load (9 nN). (h) Resistance vs normal applied load. (Reproduced with permission from Ref. [55]).
topography image [Fig. 5(a)], the resistance image [Fig. 5(b)] gives more information, where three distinct regions can be identified (the high resistance PAP region and two graphene regions with different resistances). According to the corresponding resistance profile given in Fig. 5(e), the average $R$ for region I is $\sim 5 \times 10^7 \, \Omega$ and region II is $\sim 2 \times 10^6 \, \Omega$. If there are many graphene layers, it will be stiffer that can increase the contact area, thus the measured $R$ is smaller than if there are few layers. Accordingly, the high $R$ region I has corresponded to having smaller number of layers and the low $R$ region II has corresponded to having large number of layers. However, for different samples that have been analyzed involving different number of layers, the $R$ values varied between $10^5 \, \Omega$ and $10^{10} \, \Omega$ that may have risen as a result of the difference between the bonding among adjacent layers and with the underlying substrate. The friction image obtained by LFM also does not show a significant difference between the two graphene regions except at the boundary [Fig. 5(c), (f)]. As roughly determined by the height profile [Fig. 5(d)], taken from the topography image, there are less than 10 graphene layers, which is possibly the reason for this uniformity of the friction image. However, the presence of few graphene layers was further supported by taking I-V measurements at the region I under different applied loads. Figure 5(g) shows the I-V curve recorded under 9 nN load on few-layer graphene sheets (region I), which gives a $R$ value of $10^7 \, \Omega$. The $R$ value has decreased to $10^4 \, \Omega$ with the increase of applied load [Fig. 5(h)] as it is inversely proportional to the contact area, which is proportional to the applied load. By repeating the measurement for samples with different thicknesses, it has confirmed that this kind of trend in $R$ vs. load was only observed on very few-layer graphene films. Not only ME or CVD-graphene but also epitaxial graphene (EG) shows differences in current flow depending on the number of graphene layers and the substrate features as validated by Giannazzo et al. [42]. According to the extracted conductance maps obtained by current maps via C-AFM, compared to monolayer (1L) graphene, bilayer (2L) and trilayer graphene show higher conductance in steps and the conductance at the 1L/2L junction is the lowest (sample bias 30 mV). This has been attributed to the weak wave-function coupling between 1 and 2L bands and it is independent of the interaction with the substrate. Moreover, a decrease in the local resistance of 1L graphene with the decreasing step height of the SiC substrate has been observed, which relates to the interaction between graphene and the sidewalls of the SiC steps and charge transfer.

Fabrication of RGO-based conductive films to be used in electronic devices is one of the growing research fields and C-AFM has been employed in characterizing the local conductive properties of these films as well. The first report on this was published by Mativetsky et al. in 2010 showing the measurement of uniform currents over large areas of few-layer RGO films supporting the use of RGO for transparent electrodes [48]. Few layer RGO films (thickness of 3.7 nm) were obtained by annealing a GO film, deposited on SiO$_2$/Si substrate, at 750 °C in an inert atmosphere. Multilayer regions have been found to be more conductive than a monolayer (100 mV bias applied to the sample). Again, the lower conductivity in the monolayer is due to the interaction with the substrate. Also, in multilayers, there are many percolation pathways around defects, unlike in a monolayer. This was further clarified by performing C-AFM on a single RGO flake having folded areas, as depicted in Fig. 6(a). The corresponding current image and the I-V curve given in Fig. 6(b) and (c), respectively, depict that the folded areas are $3.0 \pm 0.4$ times more conductive than the flat monolayer area. It should be mentioned here that the authors have not specifically defined what are the thinner wrinkle-like lines appearing in the height image. In contrast to the low conductive wrinkles seen in CVD-graphene, these structures on the RGO flakes are more conductive than the monolayer area. However, the small round features and small local variations that appear on the monolayer

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Figure 6: (a) Topographic and (b) current (at 0.1 V sample voltage) images of a 1L RGO sheet including folded regions. Folded regions show a higher current than the 1L areas. The Au electrode is on the left side, which is out of view. (c) I-V curves measured at points 1 and 2 marked in (a) and (b). The inset shows the same data on a linear scale. Here, Layer 1 indicates a folded area (Reproduced with permission from Ref. [48]).
area have been assigned to defects created upon thermal annealing. Current imaging of continuous GO films on SiO₂/Si substrates reduced by CH₄ plasma was reported by Morikuni et al., where the individual flakes can be clearly seen in both height and current maps [56]. This indicates the smooth current flow through the interconnected RGO flakes in the monolayer film. In another report, Ye et al. showed a current distribution of a multilayer GO film deposited on a copper substrate via the electrodeposition method followed by a mild annealing (200 °C) step [57]. As expected, the RGO film displayed a lower current compared to the conductive copper substrate. The I-V curves obtained over a voltage range from -1 V to +1 V showed that the contact resistance between the doped diamond-coated AFM probe and copper substrate and probe and RGO film is about 38 kΩ and 207 kΩ, respectively. With the increase of the film thickness, the resistance has increased as a result of the non-uniformity of the film and incomplete reduction [57].

### C-AFM-assisted nanolithography of graphene

As an attempt to modify the electronic properties of graphene to fabricate graphene-based nanoelectronic devices, nanoscale patterning or nanolithography has been explored using AFM [58, 59]. During nanolithography, one can modify the surface either by removing the material or by adding additional materials by applying appropriate forces or electric fields [38]. The interest in nanolithography of graphene is enhanced as sub-10 nm graphene nanoribbons (GNRs) have been shown to behave as semiconductors due to quantum confinement and edge effect opening a transport gap [60]. Nanoscale hydrogenation and oxidation of ME-graphene, supported on SiO₂/Si substrate, via AFM nanolithography under ambient conditions was reported by Byun et al. [61]. Accordingly, at negative lithography, H⁺ ions that are electrically dissociated from water can be adsorbed on the graphene surface (cathode). These H⁺ ions reduced at the surface forming hydrogen atoms. The active hydrogen atoms and remaining H⁺ ions assist in the hydrogenation of graphene. On the other hand, at positive lithography, negatively charged oxygen-containing radicals (e.g., OH⁻ and O₂⁻) can be attached to graphene surface (anode), resulting in oxidation of graphene. AFM-assisted nanolithography depends on several parameters such as bias voltage, scanning speed, and relative humidity (RH). Figure 7(a)–(d) depicts the friction force microscopy (FFM) (left side) and topographic (right side) images demonstrating the effect of bias voltage and scan speed in patterning graphene via hydrogenation and oxidation performed at a constant loading force of 1 nN at a RH of 33% and 20%, respectively. The FFM images were employed to distinguish the patterned lines since the hydrogenated patterns are not clearly visualized in topographic images. The line width of the patterns can be controlled by changing the bias voltage and scan speed at both voltage polarities. The line width increases with tip–sample voltage (-6 to -10 V and 4.5 to 10 V) and it decreases with increasing scan speed (0.1 to 1.0 μm/s). These patterns made by both polarities are found to be insulating and have larger friction values than pristine graphene. Hydrogenation and oxidation were further confirmed by Raman spectroscopy as given in Fig. 7(e) and (f), respectively. Raman spectrum of pristine graphene has prominent G (~ 1580 cm⁻¹) and G’ (~ 2700 cm⁻¹) bands and no D band (~ 1350 cm⁻¹). Positive lithography results in a Raman spectrum similar to GO with broad D and G bands and a less prominent G’-band, indicating that the graphene lattice has been distorted upon oxidation. As explained earlier, the negatively charged oxygen radicals produced by the decomposition of water molecules facilitate the formation of oxides on either side of the graphene sheets distorting the perfect 2D graphitic lattice. Since D band is associated with breathing mode vibrations of sp²-hybridized six-membered C-rings with defects, the broad D band appeared upon positive lithography proves the defect formation. Moreover, G band is associated with in-plane vibration of sp²-hybridized carbons in chains or rings. Therefore, the broadening of the G band can be ascribed to the damages caused to the sp²-graphene lattice [62]. Unfortunately, the kind of oxygen-containing groups that have been created on the graphene sheets has not been shown. On the other hand, negative lithography results in the evolution of defect-induced D’ (~ 1620 cm⁻¹) and (D + D’) (~ 2950 cm⁻¹) bands. Additionally, a peak at ~ 1116 cm⁻¹ was observed, which could be found in hydrogenated graphene [63]. In contrast to oxygenated graphene, the D and G bands are sharper in hydrogenated graphene. During both processes the translational symmetry of sp² C–C bonds breaks, yet the influence of these two processes has a difference resulting in D band with significantly different widths. After annealing in Ar atmosphere, in hydrogenated graphene the D band intensity decreased, and G’ band intensity increased because the C-H bonds are weaker than C-O bonds. This way one can pattern monolayer graphene having conductive and insulating regions in the desired arrangement in a nanodevice. Nanolithography has also been used to pattern GO and RGO by AFM-assisted reduction of thin films of GO that can even go to a lateral resolution as small as 4 nm [60]. AFM-assisted nanolithography is basically a thermochemical nanolithography (TCNL) process, where GO can be reduced with the help of adsorbed moisture or by using a heated AFM tip [13]. Nanolithography involving tip-induced reduction of GO films by C-AFM was first reported by Mativetsky et al. [48]. It was observed that for a 7-nm-thick GO film deposited on SiO₂/Si substrates with patterned Au electrodes, at negative sample voltages above ~ 3.6 ± 0.3 V, the current increased with increasing voltage. Additionally, the current images showed that conductive patches started to form from the Au electrodes (reduction at lateral geometry). The width of these patches depends on
the magnitude of the applied negative sample voltage. No such increment of current was observed at positive sample voltages. This phenomenon is opposite to that of pristine graphene, where at both negative and positive voltages, graphene forms insulating regions due to hydrogenation and oxidation, respectively. In the case of GO, at negative sample voltages, GO that is already
an insulator is getting reduced resulting in conductive regions. It has been shown in several reports that RH plays a vital role during AFM-assisted nanolithography of GO [64]. When the measurements are conducted under controlled RHs, a water meniscus is formed around the AFM tip and acts as a localized electrochemical environment as shown in Fig. 8(a). The tip-substrate voltage splits water in the meniscus generating reactive ionic species, which are then react with the oxygen-containing functional groups in GO according to the following equation:

\[ \text{GO} + \text{H}^+ + e^- \rightarrow \text{RGO} + n\text{H}_2\text{O}. \]

The \( \text{H}^+ \) ions are the reactive reducing species toward oxygen-containing groups. With the increase of RH, the size of the reduced spots was increased. At low RHs (< 20%), obtaining a good patterning is challenging. At high RHs (> 60%), large, reduced spots can be obtained, yet film deterioration is possible. Hence, as the ideal RH, the range of 20–40% can be considered to produce the most consistently sized patches [60].

However, other factors such as biased voltage, duration, the thickness of the GO film, and the conditions of the local film or the substrate also affect the shape and size of the reduced pattern [60]. In this paragraph, these factors will be discussed in brief. Under the same bias voltage and RH, the reduced patch size increases from monolayer to multilayer GO films. More GO layers are attribute to a higher amount of intercalated water and hence more \( \text{H}^+ \) ions are involved in reduction. A similar explanation can be given for the development of larger patches at high RHs. However, if the thickness goes above a threshold value, transportation of \( \text{H}^+ \) ions between the GO film and the conductive substrate (in vertical geometry) will be obstructed. Figure 8(b) and (c) show the topographic and current images of a monolayer GO flake on Au substrate, respectively. The arrow indicates a scratch-like feature on the substrate. On the insulating GO flakes, a few high current spots were observed (wrinkles on the height image), which were attributed to the starting points of reduction. At the scratch-like defect on the substrate, the reduced spot is the largest and follows the shape of the defect, which indicates that the substrate features can affect the local reduction. As given in Fig. 8(a) and (e), different conductive patterns of choice have been created on GO films at -4.0 V of bias voltage under a constant RH. The bright lines in Fig. 8(e) are created on a multilayer GO film at three different tip scanning rates: from 1–3, the scanning rates varied as 0.1, 0.5, and 1.0 \( \mu \text{m} \text{s}^{-1} \), respectively. The longer the tip in contact with the sample, longer will be the reduction and hence, the thicker the line resulting in higher current [Fig. 8(f)]. By carrying out nanolithography on GO films (8–15 nm thick), deposited on Au/mica and p + Si substrates, at a writing speed of 1.0 \( \mu \text{m} \text{s}^{-1} \), < 40% RH, ± 10 V tip bias voltage, and < 10 nN force, Lorenzioni

**Figure 8:** (a) Schematic of the voltage-induced reduction of GO during C-AFM. The tip-substrate voltage splits water in the meniscus generating reactive \( \text{H}^+ \) ions, which are then react with the oxygen-containing functional groups in GO. (b) Topographic and (c) current images of a GO flake on Au with a substrate defect indicated by a red arrow. (d) Current image of a monolayer of GO flake, where the letters “BU” are patterned at a 1.0 \( \mu \text{m} \text{s}^{-1} \) scanning speed. (e) Current image of three RGO lines patterned under a -4.0 V sample bias at different tip scan speeds (1 = 0.1 \( \mu \text{m} \text{s}^{-1} \), 2 = 0.5 \( \mu \text{m} \text{s}^{-1} \), and 3 = 1.0 \( \mu \text{m} \text{s}^{-1} \)) and (f) the corresponding line profiles (Reproduced with permission from Ref. [60]).
et al. observed that GO on the Si substrate gave a conductive pattern only at +10 V tip bias whereas GO on Au gave a pattern under both polarities. This is because the Au substrate behaves as a counter electrode as sufficiently as the tip itself [65]. Nevertheless, tip-induced reduction may not be suitable for very thin or monolayer GO films deposited on heavily doped Si or metal substrates since there is a possibility for the formation of electric short circuits through the junction between the conductive tip and the highly conductive substrate [66].

Surface potential, work function, and electrostatic gradient by scanning Kelvin probe microscopy (KPFM) and electrostatic force microscopy (EMF)

Surface potential of mechanically exfoliated (ME), epitaxial, and CVD-graphene

Kelvin probe force microscopy (KPFM) is another interesting AFM-based technique that enables nanoscale imaging of the surface potential of not only metals and semiconductors but also organic and biological materials [37, 67]. The KPFM measures the contact potential difference (CPD) between the conductive tip and the sample caused by the electrostatic force between them, which in turn can be used to determine the work function (WF). When the AFM cantilever and the sample are electrically connected, electrons flow from the material having the lower work function to the other until the thermodynamic equilibrium is reached (Fermi levels of the two materials will be aligned). As a result, a Vcpd will be formed between the tip and the sample. An electric force will be generated due to this Vcpd, which will be nullified by applying an external bias voltage (Vdc) with the same magnitude as the Vcpd but with opposite direction (Vdc = −Vcpd). Once the electric force is null, the WF of the sample (WFs) can be calculated when the tip work function (WFT) is known (e is the electronic charge) [68].

\[
V_{CPD} = \frac{WF_t - WF_s}{e}
\]

The beauty of KPFM is that it can provide a 2D image of the topography and the CPD simultaneously. Generally, KPFM is conducted on “lift-mode,” which consists of two tracing steps. First, the cantilever is mechanically vibrated at its resonance frequency to get the topographic image in standard tapping mode. Second, the tip is lifted to a user-defined value (known as the lift height) from the sample surface, which will then be excited by an external bias voltage. Generally, KPFM can be performed in air or in vacuum. However, to determine the absolute WF of a material, ultrahigh vacuum conditions during the KPFM measurements would be better since WF is sensitive to the surface state of the sample [69]. Adsorbed moisture, gases or other impurities, underlying substrate, and temperature can greatly affect the measurement [70]. Moreover, depending on the KPFM mode, whether it is amplitude modulation (AM-KPFM) or frequency modulation (FM-KPFM), the CPD can vary even for the same material [69, 71]. Hence, a large discrepancy for CPD values can be seen in the literature. Such variation can be seen for graphene as well, where the CPD of graphene changes not only due to the difference in the measurement mode but also depending on the origin of graphene, number of layers, defects, and surface functionalities. Table 1 represents the CPD/WF obtained via KPFM of some of the selected graphene-based materials reported in the literature.

Panchal et al. compared three modes; FM-KPFM, AM-KPFM, and EFM to obtain a reliable value for WF of epitaxial 1L and 2L graphene [71]. FM-mode is sensitive to electrostatic force gradient whereas AM-Mode is sensitive to electrostatic force. Therefore, FM-mode has a relatively high spatial resolution, yet AM-Mode has a high signal-to-noise ratio. EFM measures the electric field gradient distribution above the sample surface. However, in contrast to AM and FM-Modal where the electrical properties are directly obtained via the surface potential image, the electrical properties of the sample can be obtained via EFM amplitude and phase images. All the measurements were done at the same place of the graphene device. The CPD values obtained for 1L and 2L graphene showed a threefold increase for FM-Mode compared to AM-Mode since FM-Mode can give potential images with high spatial resolution and the same goes with EFM. Figure 9(a) shows the AFM height image of mechanically exfoliated few-layer graphene (FLG) on SiO2/Si substrate [26]. Figure 9(b) and (c) is the corresponding EFM phase images taken at −2 and +3 V tip voltages (Vtip), respectively, and the line profiles taken along the white dash line are given in Fig. 9(d). The EFM phase images show reversed contrast due to the opposite polarity of the Vtip. However, the phase shift (ΔΦ) of FLG with respect to the substrate is always negative and 2L and 5L regions have different ΔΦ values. Yet, layer-to-layer contrast cannot be seen clearly due to the instrumental noise. The ΔΦ and the surface potential can be correlated with the following equation [26],

\[
\Delta \Phi = \frac{Q}{2k} C^* (V_{tip} - V_{surface})^2
\]

where Q and k are the quality factor and the spring constant of the cantilever, respectively, C* is the second derivative of the difference between the tip and the sample capacitance as a function of the vertical distance, and Vsurface is the local electrostatic potential on the sample surface. Accordingly, the surface potential of FLG is always positive (due to hole doping) and decreases with the increase of graphene layers. Figure 9e shows the EFM phase image taken at 60 °C for multilayer epitaxial graphene (Vtip = −1.5 V), where the 1L and 3L can be clearly seen with
ΔΦ = 1° in contrast to the imaging done at ambient conditions (not shown here) [84]. This is because at elevated temperatures, the moisture layer on the sample surface will be removed and the shielding effect will be minimized. Here, IFL is the interfacial layer, which is the carbon-rich reconstructed SiC surface. Figure 9(f) and (g) illustrates the averaged EFM spectroscopy data taken for different graphene layers with respect to the biases applied to the tip and the sample (−2 to +2 V). The spectroscopy images at two biases are mirror images of each other with respect to V = 0. These spectroscopy data further confirm that the graphene layers will appear in different contrasts in EFM images where the 1L graphene shows the highest phase contrast. The gray-color area in Fig. 9(f) indicates the approximate zone where the EFM imaging is unattainable due to small biases. By proper optimization of bias voltage and the environmental conditions, the EFM technique can be used in identifying the graphene layers with desirable work functions, which are appropriate for graphene-based devices.

In multilayer graphene, the sheets are bound via van der Waals interactions resulting in different electronic properties than a single layer of a graphene sheet and beyond a critical number, multilayer graphene behaves as the bulk. By carrying out KPFM measurements, in ambient conditions, on ME-graphene deposited on SiO2/Si substrate, Lee et al. showed that depending on the number of graphene sheets, the surface potential (hence the WF) changes due to the interlayer screening effect [72]. With the increase in layer number, the surface potential showed an exponential decrease, which is ascribed to a screening of the surface potential by graphene layers. Accordingly, seven- and ten-layer graphenes show a lower WF than a monolayer graphene sheet.

Grain boundaries and wrinkles are common types of line defects found on graphene sheets that affect the electrical properties of graphene [19]. Figure 10(a) and (b) shows the height and potential images of monolayer CVD-graphene after transferred onto SiO2/Si, respectively, and Fig. 10(c) shows the corresponding height and WF profiles. The yellow, green, and blue arrows indicate grain boundary, standing collapsed wrinkles, and folded wrinkles.

### TABLE 1: Comparison of CPD and WF of graphene fabricated via different methods reported in the literature.

| Sample | V_{CPD} (mV) | WF (eV) | Measurement mode | Conditions | Ref |
|--------|--------------|---------|------------------|------------|-----|
| ME-G on SiO2/Si | 5.06 (1L) | 5.11 (1L) | EFM | Vacuum | [25] |
| ME-G on SiO2/Si | 4.99 (4L) | 4.95 (7L) | AM | Ambient | [72] |
| ME-G on SiO2/Si | 357.3 (1L) | 4.41 (1L) | FM | Vacuum | [73] |
| ME-G on SiO2/Si | 489.7 (2L) | 4.52 (2L) | AM | Ambient | [74] |
| ME-G on SiO2/Si | 614.3 (4L) | AM | Ambient and dry N2 | EFM | [75] |
| ME-G on Cu (1,2 L) | 91 (1L-2L) | 4.4 (1L) | Dual pass FM | Ambient | [76] |
| ME-G on SiO2/ (θ > 16°) | 357.3 (1L) | 4.55 (1L) | Single pass FM | Ambient | [71] |
| CVD-G on Mo After transfer to SiO2 | 150 (1L-2L) | 4.92 | AM | [77] |
| GO annealed @ 750 °C in Ar | 51 (S-1L) | 4.92 | AM | [78] |
| GO annealed @ 1000 °C in ethanol | 28 (1L-2L) | 4.92 | AM | [79] |
| ME-G on SiO2/Si | 20 (S-1L) | 4.92 | Dual pass | Ambient | [80] |
| GO | 1020 | 4.92 | Single pass | Ambient | [81] |
| GO on SiO2/Si | 20 (S-1L) | 4.92 | Single pass | Ambient | [82] |
| GO on HOPG @ 60 °C Reduction by UV | −117 (S-1L) | 4.62 | Dual pass | Ambient | [83] |
| GO on HOPG @ 450 °C | 450 | 4.56 | Dual pass | Ambient | [84] |
| GO on SiO2/Si annealed @ 800 °C in ethanol | 195 (1L-2L) | 4.56 | Dual pass FM | Ambient | [85] |

*θ = angle between two graphene grains.
respectively, which are also schematically given in Fig. 10(d). The wrinkles are thicker than the flat graphene area and standing collapsed wrinkles are much taller (1–5 nm). Although the height image does not give much information about different kinds of wrinkles, potential image clearly distinguishes the two types. The folded wrinkle appears as a single bright line and the standing collapsed wrinkle appears as a single dark line bordered by two bright lines (according to the authors, brighter contrast in potential represents lower WF). Folded wrinkles are similar to few-layer graphene due to the screening effect, and hence, the WF decreases more than the defect-free monolayer area. On the other hand, the grain boundaries show the lowest WF, but the value has varied from sample to sample, which is ascribed to the variation in stitching angle between adjacent graphene grains. Figure 10(e) and (f) depicts height and potential images of CVD-graphene deposited on SiO2/Si substrate, respectively, where the few-layer areas and wrinkles show a higher CPD (lower WF) than the monolayer areas due to the increase in Fermi level since the electron density is high [54]. Several islands are denoted in the potential map, which is separated by wrinkles. Such an inhomogeneous potential distribution is created because the wrinkles act as potential barriers for charge carriers. The drop in CPD within islands 1–3 and rise in island 4 are attributed to p-doped and n-doped nature at these islands, respectively. Islands 5–7 show a wavy nature in both height and potential maps, which leads to inhomogeneous charge distribution. To avoid the formation of large wrinkles, Vasic et al. proposed a method to synthesize graphene by a CVD method using Mo as the substrate and characterized via C-AFM and KPFM [16]. A relatively homogeneous CPD map (CPD = 352 mV, WF = 4.66 eV) over a large area was obtained with a few irregular patches due to small wrinkles. After transferring this graphene on to SiO2/Si substrate, the CPD has decreased to 205 mV (4.8 eV) because of the charge transfer from the Mo substrate, which has a lower WF than graphene.

### Surface potential of graphene oxide (GO) and reduced graphene oxide (RGO)

In contrast to ME, epitaxial, and CVD-graphene, GO and RGO have many kinds of defects that could drastically change the surface potential, particularly due to the difference in charge transfer effect associated with the oxygen-containing functional groups [83, 85, 86]. By carrying out a systematic study on the dependence of WF on the oxygen content of GO during chemical reduction, Mishra et al. have reported that the CPD value decreases (WF increases) with the removal of oxygen functionalities from GO (reduction of GO) [78]. For comparison, graphene films were fabricated via liquid phase exfoliation of graphite, and they showed the lowest CPD or highest WF than RGO films indicating the presence of fewer defects and oxygen functionalities. Figure 11(a) shows the AFM height image of 1L and 2L GO flakes on SiO2/Si substrate, and Fig. 11(b) and (c) shows the potential images taken at the same place (sample biased) before and after thermal annealing of GO at ambient conditions [83].
Figure 10: (a) AFM height and (b) potential images of 1L CVD-graphene on SiO$_2$/Si substrate. The tilt angle between the edges of the two grains is 43°. Standing collapsed wrinkles, folded wrinkles, and grain boundaries are indicated by blue, yellow, and green arrows, respectively. (c) Line profiles taken along the red line in (a) and (b) (scale bar = 500 nm). (d) Schematic illustration of folded wrinkle, standing collapsed wrinkle, and grain boundary, respectively, on graphene grown by CVD (Reproduced with permission from Ref. [19]). (e, f) Height and potential maps of CVD-graphene on SiO$_2$/Si measured by KPFM (areas 1–7 in (f) are the different islands on graphene showing variation in potential) (Reproduced with permission from Ref. [54]).

Figure 11: (a) AFM height image of GO on SiO$_2$/Si. (b, c) Potential images of GO before and after thermal annealing taken at the same place (Reproduced with permission from Ref. [83]). (d) Potential images of GO on HOPG were taken at the same place under different relative humidity (RH) and temperature conditions, which are indicated in each image. (e) The line profiles taken along the red lines in (d) (Reproduced with permission from Ref. [70]).
The GO sheets were not distinguishable, however, after annealing, the flakes were seen clearly where the 2L showed a lower potential than the 1L. GO is an insulator with many oxygen-containing groups and defects and adsorbed moisture. During thermal annealing, the oxygen functionalities will be removed resulting in conductive RGO flakes, yet the adsorbed moisture layer can affect the measurement. The layer-to-layer contrast in the potential image of RGO can be ascribed to the difference in charge distribution and screening effect since the 2L is on top of the 1L. Jaafar et al. performed KPFM measurements of GO films on Au, HOPG, and heavily p-doped Si substrates in high vacuum (HV) conditions and at 60 °C (tip biased) where the surface potential of GO decreases in steps with the increase of layer number and saturated above 5–7 layers irrespective to the type of substrate [70]. Here, the GO flakes were distinguishable in the multilayers and the steps were seen clearly in the potential images probably since the measurements were done in HV so that the shielding effect can be eliminated. As given in Fig. 11(d) and (e), the resolution of the GO flakes is highly dependent on the relative humidity and the temperature. As the RH increased, the potential steps were not distinct and the substrate to 1L surface potential decreased further proving the shielding effect from the water layer on the sample surface. Moreover, GO is hydrophilic due to the oxygen functionalities and hence, moisture layers can be easily present on the sample. Therefore, HV conditions are best for the KPFM measurements of GO films. There are many reports on measuring the CPD and hence the WF of GO and GO after reducing by different means via KPFM. However, these reports have large variations for CPD and WF that can be due to many reasons: differences in measurement mode, environmental conditions, substrate type, and amount of oxygen functionalities/defects/dopants. Hence, care should be taken when comparing the experimental results with literature, and following the best measurement conditions would minimize the ambiguity.

Electromechanical response by piezoresponse force microscopy (PFM)

The discovery of piezoelectricity in 2D materials like transition metal dichalcogenides opens a possibility for them to be used in next-generation miniaturized energy harvesters, actuators, and sensors [87]. This has motivated researchers to fabricate graphene-based piezoelectric devices. Pristine graphene is not intrinsically piezoelectric as it exhibits inversion symmetry. However, piezoelectricity has been observed in chemically doped graphene, defected graphene, twisted bilayer graphene, and graphene-based van der Waals heterostructures [88, 89], where a strain and a bandgap opening will be induced. Depending on the type of dopant atoms, their concentration, and the position of doping the in-plane strain created on the graphene lattice varied resulting in different piezoelectric coefficients. Ong and Reed theoretically showed that non-piezoelectric graphene can be transformed into a piezoelectric material and the maximum piezoelectric coefficient, \( d_{31} \), was 0.3 pm/V for graphene doped with fluorine and lithium, which is comparable to known values of other 2D and 3D materials (e.g., 0.33 pm/V for wurtzite boron nitride) [88]. Proof for piezoelectricity in graphene has also been shown experimentally and nanoscale characterization of piezoelectricity via piezoresponse force microscopy (PFM) has been conducted [27]. Although there are not many reports on the characterization of nanomechanical properties of graphene via PFM, in this section, we try to present details as much as possible as a support for the review since it is another AFM-based characterization technique. The PFM is an AFM technique that measures the local electromechanical response of a material, where an AC voltage is applied between the tip and the sample that induces a localized strain. Two important pieces of information can be obtained from the amplitude and the phase images: the piezoelectric coefficient and the polarization direction, respectively. Generally, PFM is carried out in contact mode.

In addition to the chemical doping of graphene, a strain is generated by the underlying substrate where graphene is grown or deposited due to the chemical and/or van der Waals interactions, which can affect the electromechanical properties. Figure 12(a) schematically shows the arrangement when a monolayer of CVD-grown graphene is transferred onto a SiO₂/Si calibration grating substrate with rectangular grooves (1.317 ± 10 nm) [27]. Graphene is suspended at the grooves and supported on the substrate, where the suspended areas of graphene show a higher topography due to the attractive forces exerted by the tip. The PFM measurements carried out at 90 kHz and 5 V tip bias showed that both supported and suspended graphene areas have negative piezoresponse with respect to the zero PFM signal of the substrate and the PFM signal on the supported graphene is four times higher. Moreover, by carrying out piezoresponse voltage spectroscopy measurement (-10 V to + 10 V d.c. bias), it was found that at negative tip voltages, the amplitude of the PFM signal is increasing and vice versa. The supported graphene has a net dipole moment and a polarization due to the chemical interaction of graphene with SiO₂ substrate (C⁺ and O⁻) [Fig. 12(c)]. Under negative tip bias, the dipoles will orient and increase the net polarization. In contrast, under positive bias, the dipoles will deviate from the normal and the net polarization will decrease. As a result, the PFM amplitude signal increases or decreases. The calculated piezoelectric coefficient, \( d_{33} \), of this graphene is 1.4 nm/V. Such a trend was not shown by the suspended graphene for the same voltage range since there is no chemical interaction. In a recent work, McGilly et al. showed an astonishing piezoresponse in several twisted van der Waals heterostructures including twisted bilayer graphene (tBLG), h-BN, and TMDCs using PFM [89]. The lateral-mode PFM amplitude and phase images of tBLG revealed that the moiré
superlattice and large piezoresponses were observed at domain walls and AA stacking areas, while AB stacking area showed a relatively smaller response. This makes PFM not only useful in determining the piezoelectricity of 2D materials but also useful in identifying moiré patterns. However, PFM can give responses that seem to be piezoelectric or ferroelectric, which can also be electrostatic, electrochemical effects, or even cantilever dynamics. Thus, careful analysis is required.

**Conclusion and perspectives**

Characterization of electrical properties of graphene-based materials at the nanoscale is growing rapidly since it is vital to understand the behavior of materials, in-ddetails, to be applied in nanoelectronic devices. This is where scanning probe microscopic (SPM) techniques such as atomic force microscopy (AFM) and scanning tunneling microscopy (STM) come into play. AFM-based electrical characterization methods have been widely employed to study not only for graphene but also for other organic or biological materials. Among which conductive AFM (C-AFM) and Kelvin probe force microscopy (KPFM) are at the top, where one can measure the local current distribution and surface potential of a material, respectively. Graphene is fabricated via different methods such as mechanical exfoliation (ME), chemical vapor deposition (CVD), epitaxial growth, and reduction of graphene oxide (GO). Depending on the fabrication method and deposited substrate, the structural properties of graphene may vary that in turn be responsible for changes in electrical and mechanical properties. Generally, ME and CVD-graphenes are known to be high-quality graphenes with a minimum amount of defects but GO is known to have an abundant defect that even after reduction cannot be completely amended. Moreover, wrinkles, grain boundaries, crystal domains, multilayers, folded edges, cracks, adsorbed molecules, and dopants are some of the unavoidable nanoscale structural imperfections that can be present even on ME and CVD-graphene sheets. These defects can critically affect the electrical properties of graphene and thus the overall performance of fabricated devices. SPM techniques can easily probe the defects and hence can judge what kind of defects are responsible for particular variation either in the current (as can be determined by C-AFM) or in the potential (as can be determined by KPFM) within the device.

C-AFM offers a way for simultaneous imaging of nanostructures and the current of graphene sheets. Defects on the film hinder the smooth current flow, which can be easily...
detected by the current image because the topographic image alone cannot give all the information. C-AFM imaging can be done either in vertical geometry or in horizontal geometry depending on the conductive nature of the substrate. It is believed that the grain boundaries and wrinkles reduce the current on graphene; hence, optimization of fabrication methods to avoid such defects is mandatory. Since C-AFM is performed in contact mode, damages to the graphene sheets by the AFM tip are expected. But it is of course can be controlled by selecting suitable measurement parameters, like load on the tip and scanning rate. C-AFM has been used to locally reduce GO and pattern graphene to fabricate p-n junctions via the nanolithography approach, which is another interesting phenomenon. KPFM is another interesting technique that can be used to determine the surface potential and hence the work function (WF) of materials. Variation of the surface potential in presence of wrinkles, grain boundaries, and multilayers can easily detect by analyzing the potential images. Substrate in which graphene is deposited affects the surface potential due to the change in charge transfer effect. Moreover, environmental conditions, particularly relative humidity and temperature, have a great influence on the WF due to the screening effect from adsorbed moisture layer. Hence, performance of KPFM measurement at vacuum conditions is highly recommended.

Defects are not always bad for the performance of devices. Some defects can improve the electrical properties of graphene such as dopants. These dopants can be either atoms like nitrogen, boron, phosphorous, or molecules. Identification of the position of these dopants solely from topography is not an easy task, although STM measurements can still be employed. Thus, nanoscale current and potential imaging are very helpful at any scenario. However, it should be stressed that selection of suitable measurement parameters and environmental conditions are highly necessary since these nanoscale measurements are sensitive to any slight variation of such conditions.

**Author contribution**

KKHDS contributed to conceptualization and writing—original draft. HHH contributed to writing—review & editing. Pamarti Viswanath contributed to review & editing. RJ and MY contributed to supervision, writing—review & editing.

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**Data availability**

Not applicable.

**Declarations**

**Conflict of interest** The authors declare that they have no known competing financial interests or personal relationships that could have appeared to influence the work reported in this paper.

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