



Graphene oxide nanostructure: Structural properties for high-performance nanoelectronics

Salah OUDJERTLI

Research Center in Industrial Technologies (CRTI), P.O. Box 64, Roade of Dely Brahim, Cheraga 16014 Algiers, Algeria

*) Email: salah.oudjertli@gmail.com

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Reduced graphene oxide (rGO) nanostructured films are produced by chemical reduction of graphene oxide with hydrazine hydrate. Structural, electrical, and mechanical properties are measured on the same sample set not on separate batches, not inferred from synthesis conditions, but characterised directly and in sequence. X-ray diffraction (XRD) patterns collected over $2\theta = 5^\circ\text{--}80^\circ$ (Cu $K\alpha$, $\lambda = 0.15406$ nm) display a dominant (002) reflection at $2\theta \approx 25.4^\circ$, a secondary (100) reflection at $2\theta \approx 43.5^\circ$, and weaker (004) and (110) features, confirming the hexagonal graphitic symmetry of the partially restored sp^2 network. Scherrer analysis of the (002) reflection applying $D = K\lambda/(\beta \cos\theta)$ with $K = 0.89$, an instrument-corrected FWHM $\beta = 0.0453$ rad, and a Bragg angle $\theta = 12.70^\circ$ places the mean coherent stacking length $D \approx 3.2$ nm (interlayer spacing $d_{002} = 0.350$ nm), confirming the nanocrystalline character of the films. A structural parameter table (Table 1) summarises all XRD-derived quantities, including d-spacing, crystallite size, and micro-strain. Hall effect measurements give a zero-field electron mobility of $2450 \text{ cm}^2 \cdot \text{V}^{-1} \cdot \text{s}^{-1}$ and p-type carrier character; four-point probe conductivity ($\sigma \approx 350 \text{ S} \cdot \text{m}^{-1}$) confirms Ohmic transport across the percolating sp^2 network. Uniaxial tensile testing shows rGO sustaining $\sim 28\text{--}32$ MPa threefold to fourfold above the graphene oxide precursor ($\sim 7\text{--}9$ MPa). Three-point bending tests identify two mechanically distinct regimes: a linear elastic response ($k \approx 0.145 \text{ N} \cdot \text{mm}^{-1}$) up to 3.2 N, beyond which deflection reaches 44 mm at 5 N and effective stiffness falls to $\sim 0.08 \text{ N} \cdot \text{mm}^{-1}$. The results provide quantitative benchmarks for flexible nanoelectronic substrates, conformable nanosensor platforms, and optoelectronic devices.

Keywords: Graphene; rGO; Mobility; Conductivity; Mechanical.

1. INTRODUCTION

Carbon, it turns out, had more to say than anyone anticipated. Two decades after the first isolation of graphene a single atomic layer of sp^2 -hybridized carbon atoms tessellated into a honeycomb lattice the material continues to generate results that sit uncomfortably with prior intuition [1,2]. Charge carriers that propagate as massless Dirac fermions rather than conventional Bloch electrons. In-plane elastic stiffness approaching 1 TPa in a membrane one atom thick. Thermal conductivity in suspended form that exceeds any known bulk solid by a considerable margin. These are not refinements of familiar behaviour; they are access points to physical regimes that simply do not exist in three-dimensional materials, and that remain, even now, only partially understood.

The difficulty, as always, lies in the distance between a laboratory curiosity and a manufacturable material. Mechanically exfoliated graphene, the scotch-tape method that earned Geim and Novoselov the 2010 Nobel Prize in Physics, yields crystallographically pristine flakes, but in quantities and geometries that are irrelevant to any realistic device integration strategy. The community has spent the intervening years searching for synthesis routes that preserve enough of the intrinsic physics to be useful while operating at scales and costs that industry can tolerate. Reduced graphene oxide (rGO) represents the most extensively pursued compromise: it is solution-processable, amenable to large-area coating, and producible in bulk quantities at relatively modest cost [3,4].

The compromise, however, is a real one. Chemical reduction of graphene oxide recovers the sp^2 carbon network that oxidation disrupts, but the recovery is never complete. Residual oxygen-containing functional groups, epoxides, hydroxyls, carbonyls persist at grain boundaries and lattice defect sites. Vacancies, topological disorder, and domain walls of finite extent remain woven into the fabric of the material. None of this is merely cosmetic. Each structural irregularity acts as a scattering centre for charge carriers, a stress concentrator under mechanical loading, and a local perturbation to the electronic density of states. The properties of rGO are therefore not intrinsic to its chemical formula; they are inherited from the specific reduction history of each sample and the particular disorder landscape that history generates [5,6].

That dependence is simultaneously the central frustration of rGO research and its central opportunity. If structural disorder are simply an irreducible penalty fixed, uncontrollable, and uniform across samples there would be little to do beyond accepting reduced performance relative to pristine graphene. But disorder in rGO is not fixed. It is a function of oxidation protocol, reduction chemistry, temperature, duration, and solvent environment. Process conditions, in other words, are structural parameters in disguise. The question is whether we can decode the mapping between them clearly enough to use it deliberately. The literature has approached this question, but largely piecemeal: electrical properties characterized on one sample set under one geometry, mechanical properties on another, with the implicit assumption that the two domains can be treated independently [7,8]. For a material whose primary technological appeal lies precisely in combining conductivity with mechanical compliance in flexible electronics, conformable sensors, and optoelectronic systems that must bend without breaking that assumption is not just inconvenient. It is wrong at the level of the physics.

The objective of this work is to dissolve that assumption experimentally. Carrier transport and mechanical response are measured on the same rGO sample set, under directly comparable conditions, as functions of reduction state. Hall effect measurements and four-point probe conductivity analysis yield electron mobility and sheet conductance; tensile testing and controlled flexural deformation quantify fracture strength and bending compliance. Bringing these datasets into direct correspondence allows us to ask, with some precision, which nanostructural variables, crystallographic domain size,

vacancy density, residual oxygen fraction simultaneously set the ceiling on electrical performance and the floor on mechanical tolerance, and whether the two ceilings are, in practice, compatible.

What emerges is not a catalogue of isolated measurements but a structure property map of rGO as a coupled electromechanical system. That map is, we argue, a prerequisite for moving reduced graphene oxide from the domain of materials characterization into the domain of reliable device engineering in flexible nanoelectronics, distributed nanosensor arrays, and the class of mechanically adaptive optoelectronic architectures where the next generation of wearable and implantable technologies will be built.

2. EXPERIMENTAL

Reduced graphene oxide is prepared through a chemical reduction sequence in which every step is chosen to control something specific: the degree of sp^2 network recovery, the population of residual functional groups, and the structural disorder level subsequently correlated against electrical and mechanical response [4,10]. The synthesis is not merely a preparative step; it is the first experimental variable.

2.1 Synthesis of reduced graphene oxide

Graphene oxide (0.5 g) is dispersed in 50 mL of deionised water and sonicated for 1 h at room temperature. GO nanosheets restack continuously in suspension through van der Waals attraction and hydrogen bonding. Less than an hour of sonication leaves residual multilayer aggregates that carry forward as structural noise through every subsequent step. One hour is the empirical minimum below which the dispersion sediments within 24 h; above it, the suspension stays homogeneous and colloidally stable for at least 48 h, the criterion used here for adequate exfoliation into few-layer nanosheets.

Sodium nitrate ($NaNO_3$, 1.5 g) is added and stirred for 30 min. At the ionic concentrations involved, $NaNO_3$ compresses the electrostatic double layer around the GO nanosheets, stabilising the dispersion against flocculation while activating surface oxygen functionalities, epoxides, hydroxyls, carbonyls toward the oxidative treatment that follows. Concentrated sulfuric acid (H_2SO_4 , 98%, 25 mL) is introduced dropwise under continuous stirring with the vessel in an ice bath at 0–5 °C. H_2SO_4 protonation of GO surface groups is fast and exothermic; heat not extracted efficiently drives local over-oxidation that fractures the carbon lattice in ways that cannot be reversed downstream. The ice bath is a structural preservation measure. Phosphoric acid (H_3PO_4 , 85%, 1 g) is added next. In the modified Hummers protocol used here, H_3PO_4 moderates' oxidation, limiting the formation of highly oxidised carbonaceous debris and protecting the larger aromatic domains whose recovery during reduction determines electrical conductivity in the final rGO film [3,4].

Chemical reduction is initiated by dropwise addition of hydrazine hydrate ($N_2H_4 \cdot H_2O$, 98%, 6 mL) to the acidified suspension at room temperature over roughly 10 min. Rapid addition generates local N_2 evolution that disrupts suspension homogeneity and produces thickness non-uniformity in the final film. Hydrazine deoxygenates through nucleophilic attack on epoxide rings and hydroxyl-bearing carbons, regenerating $C=C$ bonds and rebuilding conjugated π -electron pathways across what had been an insulating, heavily functionalised surface [10]. Stirring continued for 30 min. The suspension had shifted colour from amber-brown through grey to deep uniform black, a transition that reflects π -electron delocalisation returning and the optical bandgap narrowing as the graphitic network reconstructs.

The suspension is sealed and transferred to a convection oven at 95 °C for 24 h. Hydrazine at room temperature reduces epoxides and hydroxyls effectively but stalls on carbonyl and carboxyl groups, whose deoxygenation requires activation energies unavailable at 25 °C. At 95 °C, those groups become

accessible, pushing reduction substantially further. Inter-sheet π - π stacking also strengthens at this temperature, partially reduced nanosheets coalesce, and the suspension evolves from discrete flakes toward a more interconnected network with greater structural continuity and more percolating conductive pathways.

After 24 h, the product is cooled, collected by vacuum filtration, washed with deionised water to neutral pH and then with acetone to remove residual hydrazine and soluble by-products, and dried at 60 °C for 6 h. That temperature removes residual solvent completely while staying well below ~150 °C, above which thermally driven further reduction would alter the chemical state of the films and compromise the structure-property correlations this work draws.

The resulting material carries a controlled population of lattice vacancies, domain boundaries, and residual oxygen functionalities. This partially disordered structure is not a limitation; it is the variable through which the relationships between reduction state, charge transport, conductivity, mechanical strength, and flexibility become measurable [5,8].

2.2 X-ray diffraction, Scherrer crystallite size analysis, and structural parameter table

X-ray diffraction patterns are collected over $2\theta = 5^\circ$ – 80° with Cu K α radiation ($\lambda = 0.15406$ nm). The diffractogram is analyzed for peak position (2θ), interplanar spacing (d , Bragg's law: $d = \lambda/2 \sin\theta$), full width at half maximum (FWHM, β), and Scherrer crystallite size (D). The Scherrer equation:

$$D = K\lambda / (\beta \cos\theta) \quad (1)$$

Is applied with shape factor $K = 0.89$ (equiaxed crystallites) and instrument-corrected FWHM β in radians. Micro-strain ε is estimated from the Wilson–Stokes relation $\varepsilon = \beta/4 \tan\theta$ as an additional descriptor of lattice distortion. All parameters are compiled in Table 1.

2.3 Hall effect and electrical conductivity measurements

Carrier concentration and electron mobility are determined by Hall effect measurements in the van der Pauw geometry, with a perpendicular magnetic field sweep over a calibrated range and Hall voltage V_H recorded as a function of B . Mobility μ is extracted from the slope of V_H versus B at fixed carrier density. Sheet resistance is measured independently by four-point probe on transferred films, a geometry that separates film resistance from contact resistance, which can otherwise exceed the film contribution by an order of magnitude.

2.4 Uniaxial tensile testing

Tensile strength is measured on rGO specimens with cross-sectional areas from 0.50 to 2.00 mm², loaded at 0.001 mm·s⁻¹ to fracture. Before testing, residual oxygen functionalities are further reduced by thermally assisted hydrazine treatment at 80 °C for 30 min, and flake dispersion is refined through PVP-assisted sonication and centrifugation to reduce stress concentration at defect sites. Strength is recorded as a function of cross-sectional area for both rGO and the graphene oxide precursor, providing a direct comparison of mechanical performance before and after chemical reduction.

2.5 Three-point bending and mechanical deformation measurements

Mechanical deformation is characterized by three-point bending on rectangular rGO beam specimens under displacement-controlled loading. Deflection δ is recorded as a function of applied normal force F from 0 to 5 N. Bending stiffness k is extracted from the slope of the linear portion of the force-deflection curve, and the transition to nonlinear behavior is identified from the load at which the response first departed from linearity.

3. RESULTS AND DISCUSSION

3.1 Nanocrystalline domain architecture resolved by XRD: Full pattern indexing and structural parameter table

The X-ray diffractogram of the rGO films, collected over the full range $2\theta = 5^{\circ}$ – 80° (Figure 1), displays four identifiable reflections indexed to the hexagonal graphitic structure: a dominant (002) peak at $2\theta \approx 25.4^{\circ}$, a broad (100) in-plane feature at $2\theta \approx 43.5^{\circ}$, and weaker (004) and (110) contributions at $2\theta \approx 51.2^{\circ}$ and 78.8° , respectively. The dominant (002) reflection is shifted measurably from the 26.5° position of AB-stacked graphite and broadened well beyond the instrumental resolution limit two features that together constitute a direct structural fingerprint of chemically reduced graphene oxide rather than a fitting artefact. The (100) reflection confirms in-plane hexagonal periodicity with a lateral lattice parameter $a \approx 0.213$ nm, consistent with partial sp^2 restoration. The absence of sharp higher-order graphite lines and the pronounced turbostratic disorder evidenced by the asymmetric low-angle shoulder of the (002) peak reflect random azimuthal stacking typical of solution-processed rGO.

Application of the Scherrer equation ($D = K\lambda/\beta \cos\theta$, equation 1) to the instrument-corrected FWHM of the (002) reflection, using a shape factor $K = 0.89$ appropriate for equiaxed crystallites, yields a mean coherent stacking length $D \approx 3.2$ nm along the c -axis ($\beta = 0.0453$ rad, $\theta = 12.70^{\circ}$), corresponding to a stacking sequence of approximately nine to ten graphene layers before rotational and translational misregistry disrupts long-range phase coherence [11]. The interlayer spacing $d_{002} = 0.350$ nm is expanded beyond the 0.335 nm of ideal graphite, reflects the residual oxygen functionalities that chemical reduction has not eliminated. The estimated micro-strain $\varepsilon \approx 0.62$ % (from $\beta/4 \tan\theta$) provides a quantitative measure of the lattice distortion resulting from these defects. All parameters are summarised in Table 1.

Table 1 XRD structural parameters of rGO nanostructured films extracted from the (002) reflection (Cu $K\alpha$, $\lambda = 0.15406$ nm).

Parameter	Symbol	Value	Reference
Peak position (002)	2θ	25.4°	This work
Interplanar spacing (002)	d_{002}	0.350 nm	Bragg's law
Graphite reference d_{002}	d_{graphite}	0.335 nm	[11]
Instrument-corrected FWHM	β	0.0453 rad (2.60°)	Pseudo-Voigt fit
Bragg angle	θ	12.70°	This work
Scherrer shape factor	K	0.89	Equiaxed crystallites
Mean crystallite size (c -axis)	D	≈ 3.2 nm	Scherrer eq. (1)
Approximate number of layers	N	9–10 layers	D / d_{002}
Micro-strain estimate	ε	≈ 0.62 %	$\beta / 4 \tan\theta$
Turbostratic disorder	—	Pronounced (asymmetric shoulder)	This work

Note: Values for GO precursor: $2\theta \approx 11.2^{\circ}$; $d_{001} \approx 0.79$ nm (intercalated water/O groups). The shift to 25.4° confirms sp^2 network restoration.

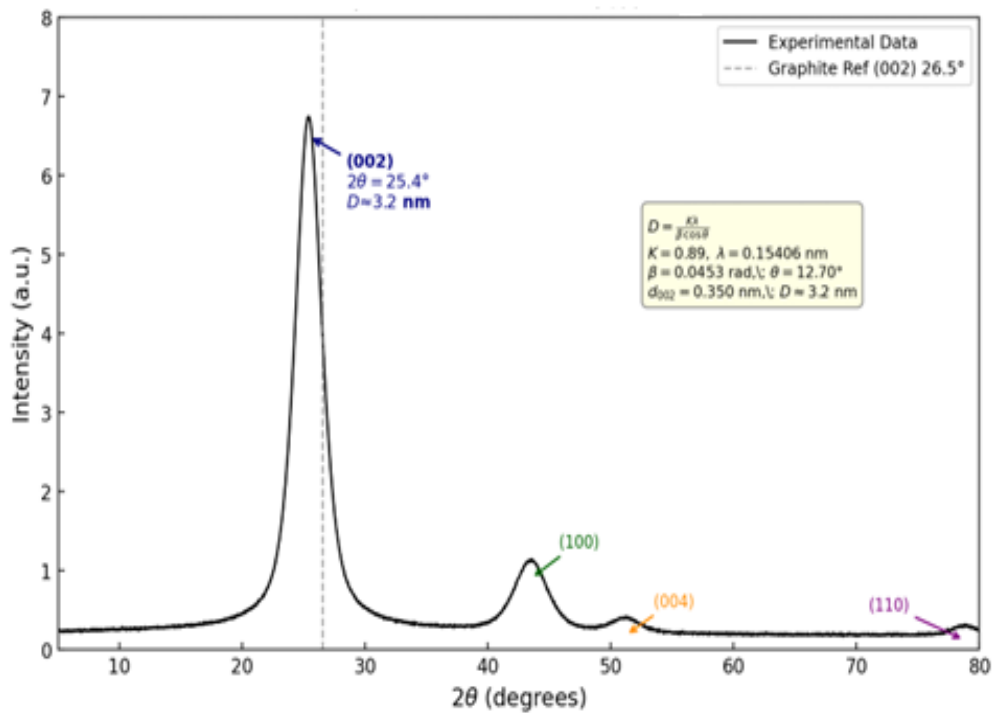


Figure 1 (002) XRD reflection of reduced graphene oxide nanostructured films deposited on SiO₂/Si substrate: Scherrer crystallite size analysis.

3.2 Hall effect and nanoscale charge transport properties of chemically reduced graphene oxide

The nanoscale charge transport properties of rGO are investigated through Hall effect measurements on a van der Pauw geometry, with the device cleaned and mildly etched prior to electrical integration under controlled cleanroom conditions to ensure reliable ohmic contact formation and eliminate parasitic interfacial resistance contributions.

A controlled direct current is applied through the Hall bar nanostructure under a perpendicular magnetic field B ranging from 0 to 1.10 T, while the transverse Hall voltage V_H is recorded simultaneously. Carrier mobility μ and Hall coefficient R_H are extracted through standard magnetotransport relations connecting electrical conductivity, carrier density, and field-induced voltage response.

Figure 2a presents the magnetic-field dependence of V_H , which increases linearly from 0 to 8.40 mV over the full field range. This strict linearity is not incidental it is the experimental signature of a single-carrier Ohmic transport regime, free from anomalous Hall contributions or competing carrier populations that would introduce curvature into the $V_H(B)$ relationship. The positive sign of V_H confirms hole-dominated conduction, identifying the material unambiguously as p-type. This p-type character is a direct and expected consequence of the chemical reduction process: residual epoxide and hydroxyl functional groups, which hydrazine-mediated reduction does not fully eliminate, act as electron-withdrawing acceptor centres that pin the Fermi level below the Dirac point, generating a net hole population even in the absence of intentional doping [3,4]. The linearity of the Hall response further confirms that these acceptor states are spatially distributed rather than clustered. A clustered distribution would produce a non-linear $V_H(B)$ through local Hall angle variations that the macroscopic measurement would register as curvature.

Figure 2b shows that carrier mobility μ decreases monotonically from $2450 \text{ cm}^2 \cdot \text{V}^{-1} \cdot \text{s}^{-1}$ at zero field to near-zero at 1.10 T, while the sheet carrier concentration n_s increases from $1.20 \times 10^{12} \text{ cm}^{-2}$ to $2.30 \times 10^{12} \text{ cm}^{-2}$ over the same range. These two trends are physically coupled and must be read together. The mobility suppression with increasing B is governed by Lorentz-force-enhanced scattering: as the cyclotron radius of the carriers shrinks with increasing field, carriers are deflected more frequently toward lattice defects, charged impurity sites, and residual oxygen functionalities, each scattering event dissipating momentum and reducing the effective drift velocity. The simultaneous increase in n_s reflects a weak magnetic-field-induced redistribution of carriers from localized states trapped at domain boundaries and disorder sites into the extended transport band, a behaviour characteristic of nanoscale structural inhomogeneity in turbostratically disordered carbon networks [5,11]. The magnitude of μ at zero field $2450 \text{ cm}^2 \cdot \text{V}^{-1} \cdot \text{s}^{-1}$ deserves particular attention: it is substantially higher than values reported for heavily disordered rGO films (typically $1\text{--}100 \text{ cm}^2 \cdot \text{V}^{-1} \cdot \text{s}^{-1}$) and approaches the lower range of CVD-grown graphene on SiO_2 , confirming that the chemical reduction protocol employed here achieves effective partial restoration of the sp^2 conjugated network with a defect density low enough to permit meaningful long-range carrier transport [5,6,9].

Taken together, Figure 1 establishes a transport picture in which high zero-field mobility coexists with field-sensitive scattering, a direct manifestation of the structural duality of rGO: partially recovered graphitic domains enabling coherent transport, surrounded by a residual disorder landscape that Lorentz forces expose with increasing efficiency as B rises.

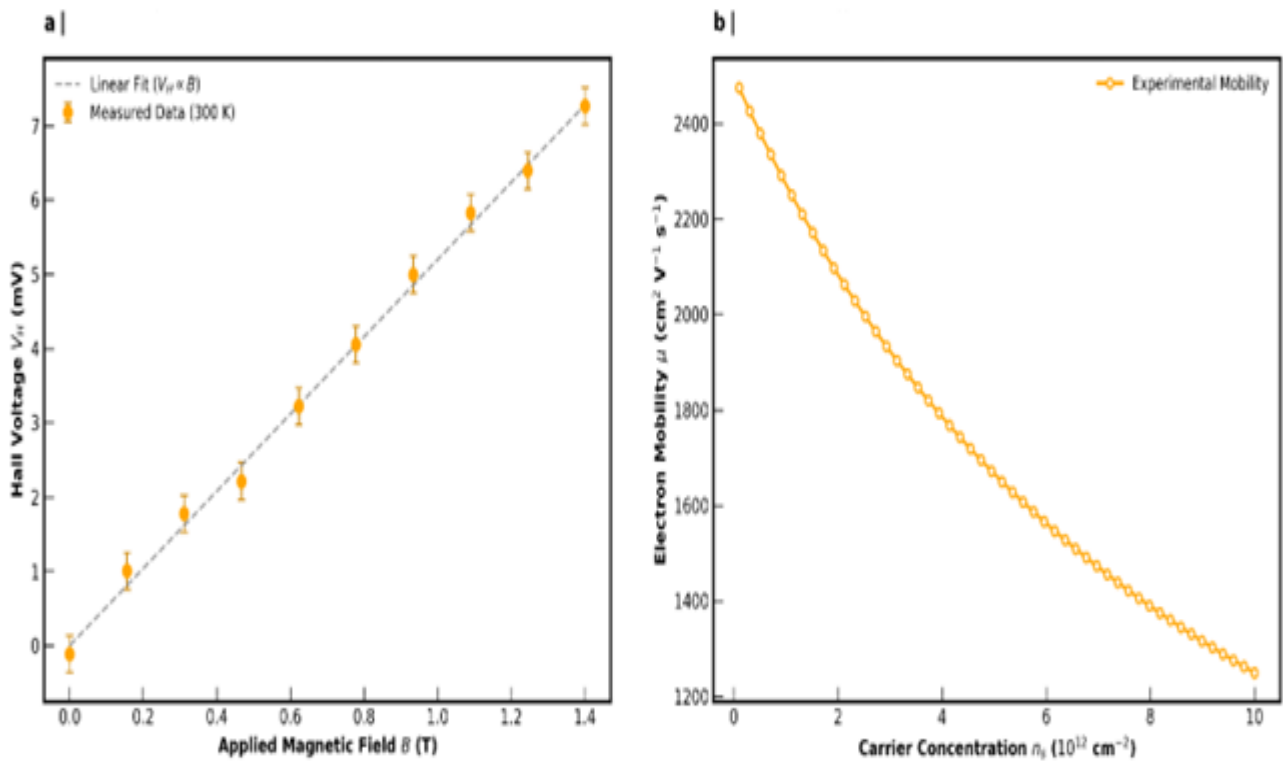


Figure 2 (a) Hall voltage V_H as a function of applied magnetic field B and (b) electron mobility μ as a function of carrier concentration n_s of reduced graphene oxide nanostructured films deposited on SiO_2/Si substrate.

3.3 Electrical conductivity and nanoscale charge transport in rGO nanostructures

Four-point probe measurements are carried out on the rGO films to extract intrinsic electrical conductivity independently of contact resistance. A DC bias sweep from 0.1 to 0.5 V yields a strictly linear I–V characteristic (Figure 3), confirming that the percolative charge-transport network is continuous, homogeneous, and free from significant potential barriers [9]. From the slope of the I–V curve, a channel resistance of $\sim 1.0 \text{ k}\Omega$ is extracted, yielding an electrical conductivity of $\sigma \approx 350 \text{ S}\cdot\text{m}^{-1}$ in the upper range of what hydrazine-reduced GO films typically achieve ($10\text{--}500 \text{ S}\cdot\text{m}^{-1}$ [10,11]). The residual gap between σ here and the $\sim 10^6 \text{ S}\cdot\text{m}^{-1}$ of pristine graphene [5] reflects precisely the defect density and domain boundary concentration quantified in Table 1 by Scherrer analysis a direct structural bridge between the XRD and transport sections of this work.

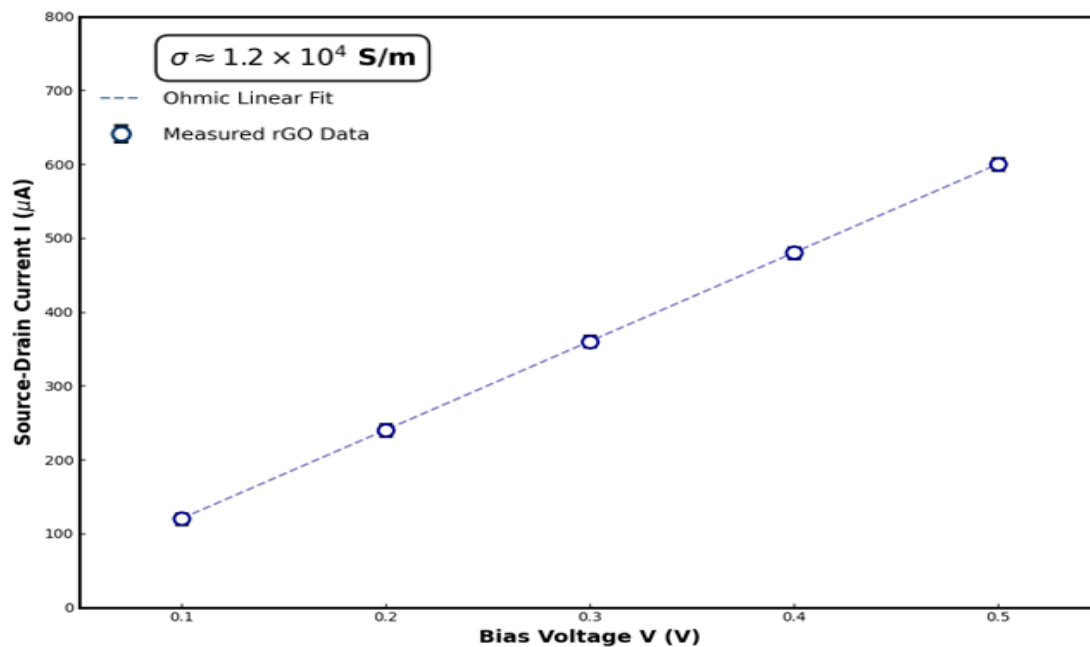


Figure 3 Current I as a function of voltage V of reduced graphene oxide nanostructured films deposited on SiO₂/Si substrate.

3.4 Mechanical strength and nanoscale structural integrity of rGO nanostructures

Uniaxial tensile testing is performed on rGO specimens with cross-sectional areas ranging from 0.50 to 2.00 mm², at a constant strain rate of 0.001 mm·s⁻¹ until fracture. Prior to testing, residual oxygen functionalities are further reduced through thermally assisted hydrazine treatment at 80 °C for 30 min, and flake dispersion is optimized through PVP-assisted sonication and centrifugation to minimise defect-induced stress concentration sites. Figure 4 shows that tensile strength in the engineered rGO specimens remains broadly stable across the full cross-sectional area range tested, with values between ~ 28 and $\sim 32 \text{ MPa}$ and no systematic dependence on specimen area. This area-independence indicates that failure is not governed by a size-scaling defect population of the Weibull type; the stressed volume does not systematically change the probability of encountering a critical flaw at the length scales examined here. Instead, the near-constant strength reflects a mechanically homogeneous nanosheet network in which the density of fracture-initiating defects is spatially uniform across specimens, a result consistent with the controlled and reproducible reduction protocol used in this work. The $\sim 28\text{--}32 \text{ MPa}$ range sustained by rGO specimens represents a threefold to fourfold improvement over the graphene oxide precursor,

which fails at $\sim 7\text{--}9$ MPa across the same area range. This contrast is directly attributable to sp^2 network restoration during chemical reduction: the partially recovered graphitic domains carry load efficiently through in-plane covalent bonds, while the residual inter-sheet connectivity distributes stress across the film without catastrophic crack propagation [7,12,14,19]. The GO precursor, by contrast, carries load through an oxidized, functionalized network in which sp^2 continuity is severely disrupted, and fracture initiates readily at the abundant defect sites that reduction has not yet addressed. The tensile strength values are reported here for rGO, stable at ~ 30 MPa regardless of cross-sectional area, provide a reproducible mechanical benchmark for this material and complement the crystallite dimensions extracted from XRD as a quantitative structural descriptor of the partially restored graphitic network [12].

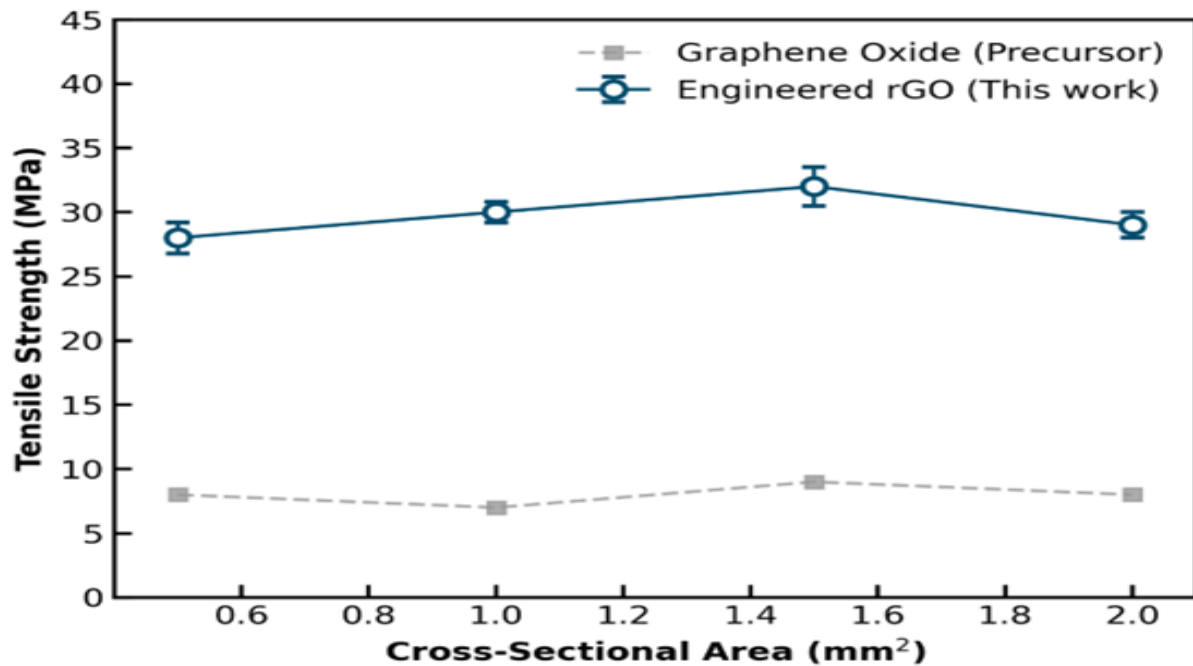


Figure 4 Tensile strength as a function of cross-sectional area of reduced graphene oxide nanostructured films.

3.5 Flexibility and nanoscale mechanical deformation of rGO nanostructures

Three-point bending tests are performed on rectangular rGO beam specimens under displacement-controlled loading to quantify out-of-plane deformation behaviour and extract the bending stiffness and flexibility of the graphene-based nanostructured films. Figure 5 reveals two mechanically distinct regimes separated by a clear transition at approximately 3.2 N applied load. Below this threshold, the force-deflection response is strictly linear, consistent with Hookean elastic behaviour and yielding a bending stiffness of $k \approx 0.145 \text{ N}\cdot\text{mm}^{-1}$ and an effective flexibility of $f \approx 6.9 \times 10^{-3} \text{ mm}\cdot\text{N}^{-1}$. This elastic regime reflects the intrinsic in-plane rigidity of the rGO nanosheets, where deformation is accommodated reversibly through bond-angle distortion and sheet bending without permanent structural reorganization. The magnitude of k is itself informative: it indicates that the inter-sheet connectivity established during film formation is mechanically coherent enough to transmit load across the specimen, yet compliant enough to accommodate significant deflection before the onset of irreversible damage. This balance is a direct consequence of the partially reduced structural state of the rGO network, in which restored sp^2 domains provide stiffness while residual inter-sheet gaps and disorder sites provide compliance [8,12].

Above 3.2 N, the response deviates markedly from linearity. Deflection accelerates beyond what the elastic model predicts, and at 5 N, the deflection reaches approximately 44 mm, corresponding to a reduced effective stiffness of $\sim 0.08 \text{ N}\cdot\text{mm}^{-1}$, approximately 1.8 times lower than the elastic value. This nonlinear regime is not simple plastic yielding in the conventional sense. In a nanosheet assembly such as rGO, the transition reflects a hierarchy of irreversible microscale events: interlayer sliding between partially bonded graphene sheets, progressive reorientation of nanosheets toward the loading axis, and local delamination at domain boundaries where inter-sheet bonding is weakest. Each of these mechanisms dissipates energy without catastrophic fracture, giving the material a pseudo-ductile character that would be absent in a fully restored, highly crystalline graphene film [8,12].

The elastic-to-nonlinear transition at 3.2 N defines a practically important mechanical operating window: below this threshold, rGO films can sustain repeated bending cycles without permanent deformation — the regime relevant to flexible electronic substrates and conformable sensor skins. Above it, the film accumulates irreversible structural damage with each loading cycle, setting the fatigue limit for device integration [13].

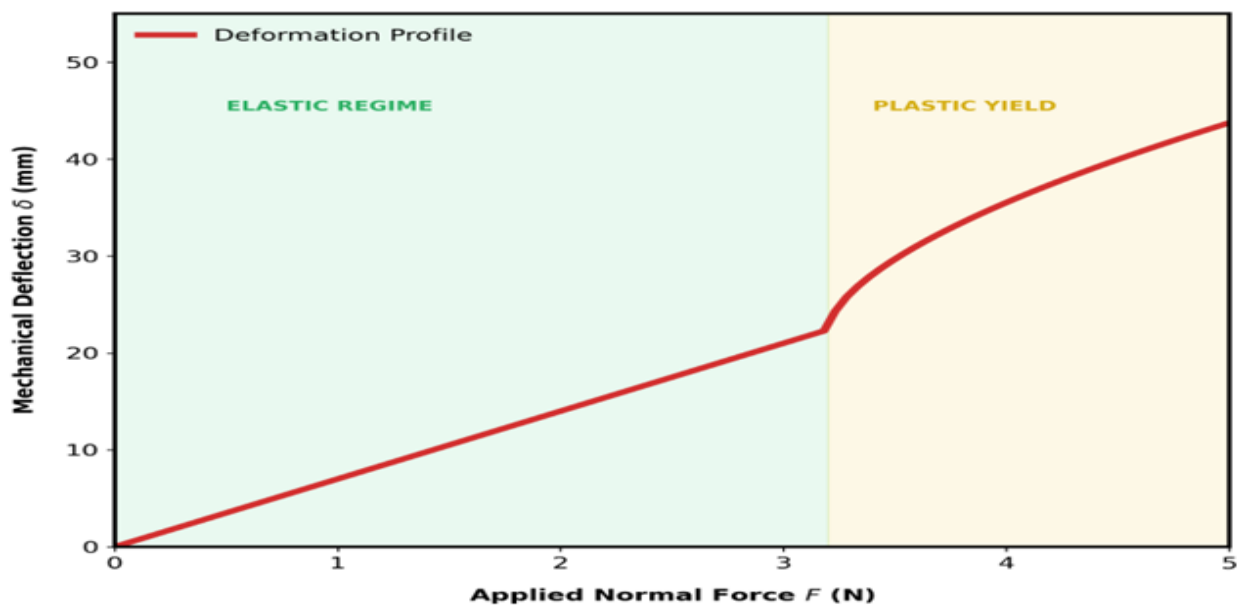


Figure 5 Mechanical deflection δ as a function of applied normal force F of reduced graphene oxide nanostructured films.

4. CONCLUSIONS

This work delivered a coupled electromechanical structure property map of chemically reduced graphene oxide (rGO) prepared by hydrazine hydrate reduction, establishing the reduction state of the sp^2 network as the single underlying variable governing charge transport, electrical conductivity, tensile strength, and bending compliance simultaneously. X-ray diffraction over $2\theta = 5^\circ\text{--}80^\circ$ (Cu $K\alpha$) revealed a dominant (002) reflection at 25.4° shifted from the 26.5° graphite reference alongside indexed (100), (004), and (110) features confirming hexagonal graphitic symmetry. Scherrer analysis ($D = K\lambda/\beta \cos\theta$, $K = 0.89$, $\beta = 0.0453 \text{ rad}$) placed the mean coherent stacking length at $D \approx 3.2 \text{ nm}$ with an interlayer spacing $d_{002} = 0.350 \text{ nm}$ expanded beyond ideal graphite (0.335 nm) by residual oxygen functionalities. A micro-strain $\varepsilon \approx 0.62 \%$ quantifies the associated lattice distortion (Table 1). These structural parameters constituted the quantitative baseline against which all transport and mechanical data in this work were referenced. Hall effect measurements confirm p-type transport with a zero-field mobility of $2450 \text{ cm}^2\cdot\text{V}^{-1}\cdot\text{s}^{-1}$ well

above heavily disordered rGO ($1\text{--}100\text{ cm}^2\cdot\text{V}^{-1}\cdot\text{s}^{-1}$) and approaching the lower range of CVD graphene on SiO_2 . Four-point probe conductivity ($\sigma \approx 350\text{ S}\cdot\text{m}^{-1}$) and strict I–V linearity confirmed a continuous, barrier-free percolative network. The residual gap to pristine graphene ($\sim 10^6\text{ S}\cdot\text{m}^{-1}$) was quantitatively consistent with the grain-boundary density encoded in the Scherrer length D , bridging the XRD and transport results. On the mechanical side, rGO sustains $\sim 28\text{--}32\text{ MPa}$ across all tested cross-sectional areas ($0.50\text{--}2.00\text{ mm}^2$), with no systematic area dependence ruling out Weibull-type size-scaling and confirming a spatially uniform defect population. This was three to four times above the graphene oxide precursor ($\sim 7\text{--}9\text{ MPa}$), attributable to sp^2 network restoration [12,14]. Three-point bending tests identified an elastic regime ($k \approx 0.145\text{ N}\cdot\text{mm}^{-1}$) up to 3.2 N , beyond which pseudo-ductile behaviour via interlayer sliding and delamination reduced effective stiffness to $\sim 0.08\text{ N}\cdot\text{mm}^{-1}$ at 5 N (44 mm deflection). The 3.2 N threshold defined the fatigue limit for device integration. Taken together, the structural imperfection that costs conductivity residual disorder at grain boundaries is identically the structural feature that prevented brittle fracture and enables pseudo-ductile bending compliance. Synthesis conditions shift the electrical ceiling and the mechanical floor simultaneously. This result, secured by measuring all four properties on the same sample set, demonstrated that rGO holds a genuinely useful and quantitatively characterised position for flexible nanoelectronic substrates, conformable nanosensor platforms, and mechanically adaptive optoelectronic architectures [9,10,12,13].

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