

1 **Enhancing mechanisms of arc-erosion resistance for copper tungsten**
2 **electrical contact using reduced graphene oxides *in situ* modified by**
3 **copper nanoparticles**

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16

17 **Abstract:** To solve critical issues of premature failure for copper tungsten (CuW) based
18 electrical contacts during arc erosion at the moment of arc breakdown, we proposed a
19 new strategy of using metal doped reduced graphene oxides (rGOs) and *in-situ* formed
20 tungsten carbides to inhibit movements of cathode spots during the arc ablation process.

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1 CuW composites were reinforced with Cu modified rGO nanopowders (i.e. Cu@rGO)
2 using combined processes of chemical co-reduction, ball milling and spark plasms
3 sintering (SPS). Effects of Cu@rGO addition on microstructure, arc erosion resistance
4 and arc ablation resistance of the CuW composites were systematically investigated.
5 Results showed that tungsten carbides with irregular shapes were formed through *in-*
6 *situ* reactions of rGO and tungsten during the SPS process. Arc erosion resistance of
7 CuW composites was significantly improved owing to introduction of nanostructured
8 Cu@rGO. Compared with those of CuW composites, the ablation areas of
9 Cu@rGO/CuW ones were much smaller and the ablation craters were shallower, and
10 the average strengths of dielectric vacuum breakdowns of the CuW composites with 3
11 wt% Cu@rGO were increased by 28.9%. The arc breakdown mechanisms of
12 Cu@rGO/CuW composites were identified as: (1) The nanostructured Cu@rGO
13 increases the viscosity of molten metal Cu, thus inhibiting its fast flow and splashing;
14 (2) Lower work functions of carbon (i.e. rGO) and tungsten carbide restrain the electron
15 emissions during arc breakdown; and (3) The tungsten carbides with their good stability
16 and high melting point shorten the solidification time of molten copper liquid and
17 extend the service life time of the Cu@rGO/CuW composites.

18

19 **Keywords:** Cu@rGO nanoparticles, microstructure, CuW composites, arc-erosion
20 resistance

21

22 **1. Introduction**

1 Electrical contacts are key components used in high-voltage circuit-breaker, and
2 are responsible for switching the currents on and off at super-high voltages. However,
3 electric arc generated during contact and break operations inevitably destroys the
4 electrode's surface under the combined actions of transient heat, mechanical impact and
5 huge electromagnetic voltage. Therefore, performance and reliability of the contact
6 materials will directly affect both life-time of switches and reliability of electrical
7 operations [1-5]. It is crucial to ensure the contact materials with good thermal/electrical
8 conductivities, high hardness, and high ablation resistance. Copper-tungsten (CuW)
9 alloys have been explored for this application, which is benefited from the exceptional
10 mechanical-physical properties of W skeleton (with its high hardness and high
11 temperature strength, but a low thermal expansion coefficient) and excellent
12 conductivity of Cu (with its good plasticity and thermal and electrical conductivities,
13 but a high thermal expansion coefficient). CuW is often regarded as a typical pseudo-
14 alloy without mutual miscibility or reactions between W and Cu [6-7].

15 Recently, there are increased requests for the CuW alloys to be applied as high
16 power/voltage electrical contacts under severe service conditions. Therefore, it is
17 critical to improve the contact materials' mechanical properties and arc ablation
18 resistance. Two main approaches have often been adopted for this purpose. The first
19 one is to refine grains of W matrix [8-10]. Frequently nano-sized W powders were
20 chosen as the raw material, however, this often results in their severe agglomeration
21 during blending/mixing process. The second commonly applied method is to introduce
22 nanoscale second phases which can provide reinforcement or functionalization

1 purposes. These materials commonly include carbon nanomaterials, rare earth and their
2 oxides, and many hard ceramic nanoparticles [11-15].

3 Graphene has been extensively explored to enhance functional and mechanical
4 properties of metal matrix nanocomposites for wide-range applications due to its
5 superior mechanical, electrical and thermal properties [16-20]. However, strengthening
6 using graphene in the graphene/Cu system often shows limited effect due to the easy
7 agglomeration of graphene in the matrix and a weak interfacial bonding between
8 graphene and Cu matrix. This weak bonding is mainly due to the large specific surface
9 areas and van der Waals force of graphene, and the poor wettability of carbon materials
10 (e.g., graphene) with metal of Cu (e.g., the contact angle between Cu and graphite is
11 $\sim 140^\circ$).

12 One of the viable approaches to improve dispersion and wettability of graphene
13 inside the metal matrix composites is to coat it with suitable nanomaterials of metals or
14 carbides [17, 21]. These metal or carbide nanoparticles/nanolayers decorated onto the
15 surfaces of carbon nanomaterials can effectively minimize the density differences
16 between graphene and metal matrix, thus leading to improved dispersion effects of
17 graphene. For example, we recently reported that a good balance of strength and
18 ductility was achieved in a Ti composite using reduced graphene oxide nanosheets
19 decorated with metal or ceramics nanoparticles [17, 21]. Chu et al. [22] also improved
20 the interfacial adhesion and tensile strength of reduced graphene oxide /CuCr
21 composites by alloying with Cr to *in-situ* form a nanolayer of Cr_7C_3 . We previously also
22 introduced Cu-coated graphene nanoplates (Cu@GNPs) into the CuW composites

1 using electroless plating and spark plasma infiltrating sintering [12]. Results showed
2 that electrical conductivity, thermal conductivity, and micro-hardness have been
3 increased up to ~ 95.3%, ~ 24.3%, and ~ 28%, respectively, compared with those from
4 the conventionally sintered CuW powders. Although the electrical and mechanical
5 properties of CuW composites reinforced with modified graphene have been
6 significantly improved, the arc-erosion behavior of CuW electrical contacts doped with
7 nanostructured Cu@rGO is still unsatisfied. A previous study [5] reported that the arc
8 breakdown position on the surface of CuW alloys occurred mainly on the rich Cu phase
9 and Cu/W interfacial zones, where liquid Cu was sputtered intensely and larger cathode
10 craters were formed, accelerating the centralization arc erosion and cracking along
11 Cu/W interface. Hence, interfacial strengthening is critical for long service life-time of
12 CuW electrical contacts composites.

13 Herein, copper nanoparticles were firstly decorated onto the reduced graphene
14 oxides nanosheets (named as Cu@rGO for simplification) and then further applied as
15 reinforcements to strengthen Cu/W interfaces for fabrication Cu@rGO/CuW
16 composites. Firstly, nanostructured Cu@rGO can effectively minimize the density
17 differences between graphene and the CuW matrix, thus leading to improved dispersion
18 effects of graphene. Secondly, the *in-situ* formed tungsten carbides can enhance the
19 interfacial bonding strength via reactive wetting on the surfaces. The Cu@rGO
20 nanopowders were fabricated using a one-step chemical co-reduction process, and then
21 incorporated into the CuW matrix to fabricate rGO/CuW composites using a powder
22 metallurgy route. The influences of Cu@rGO nanopowders on the microstructure and

1 arc erosion resistance were investigated. The key mechanisms of rGO/CuW composites
2 arc breakdown were further studied. Tungsten carbides with irregular shapes were *in-*
3 *situ* formed during sintering process. Results showed that the electrical breakdown
4 strength of 3 wt% Cu@rGO/CuW composites were increased by 28.9% in comparison
5 with those of the CuW composites. The reasons can be attributed to: (a) Addition of
6 nanostructured Cu@rGO increases the viscosity of molten metal Cu, thus inhibiting the
7 significant flow and splashing of molten metal; (b) The lower work functions of carbon
8 and tungsten carbide restrained the electron emission; and (c) Tungsten carbides with
9 their good stability and high melting point shorten the solidification time of molten
10 copper liquid, thus extending service life time of the Cu@rGO/CuW composites.

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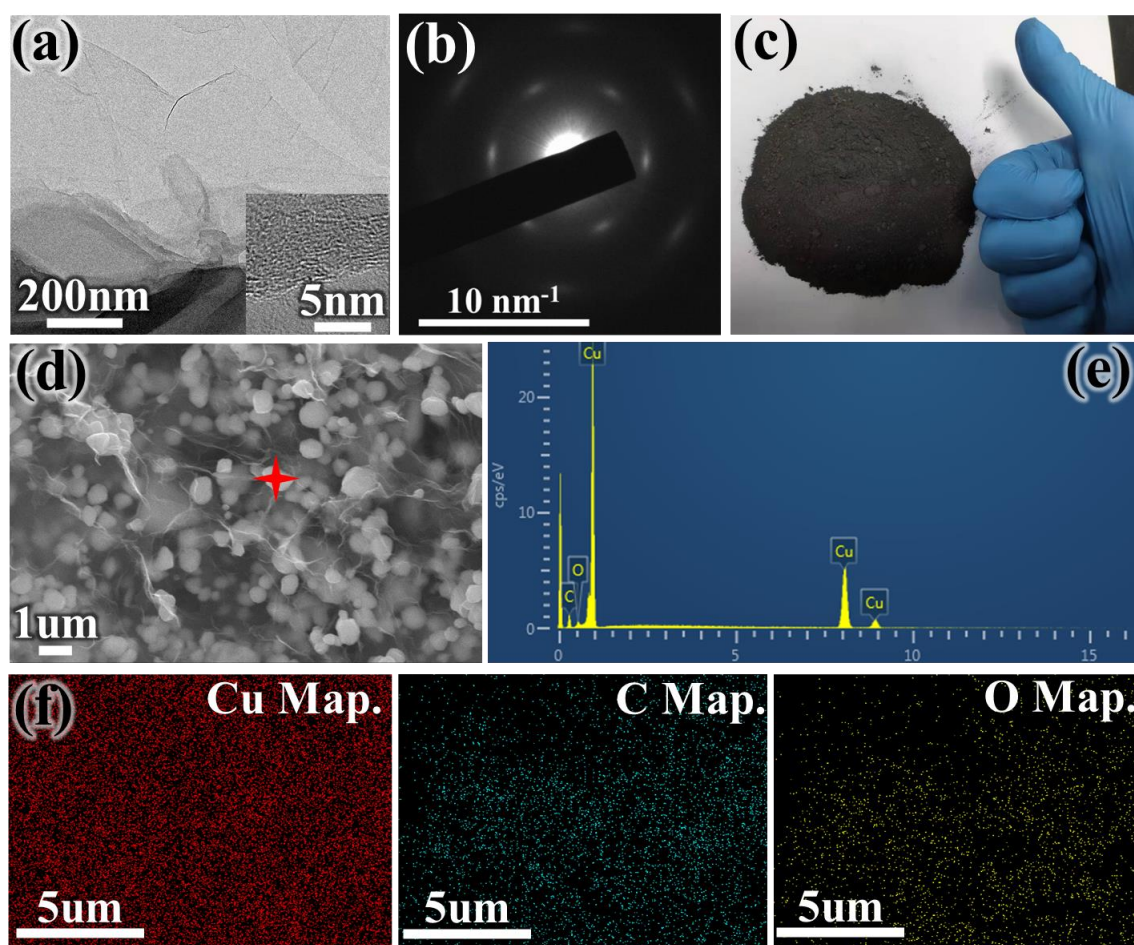
12 **2. Materials and Experimental methods**

13 **2.1 Materials and methods**

14 Graphene oxides nanosheets (GONs) prepared using the Hummers method were
15 purchased from XFNANO Technology Co., Ltd., China. A transmission electron
16 microscope (TEM) image shown in **Figure 1(a)** reveals that the GONs have large-scale,
17 transparent, and folded structures. However, there are many nanoscale defects at the
18 boundary of GONs (inset in **Figure 1a**), which could facilitate the easy *in-situ*
19 deposition of Cu nanoparticles onto the surfaces of GONs. Select electron diffraction
20 pattern of the GONs shown in **Figure 1b** confirms the low degree of crystallinity for
21 the GONs.

22 The as-synthesized GONs were reduced into the rGOs and simultaneously coated

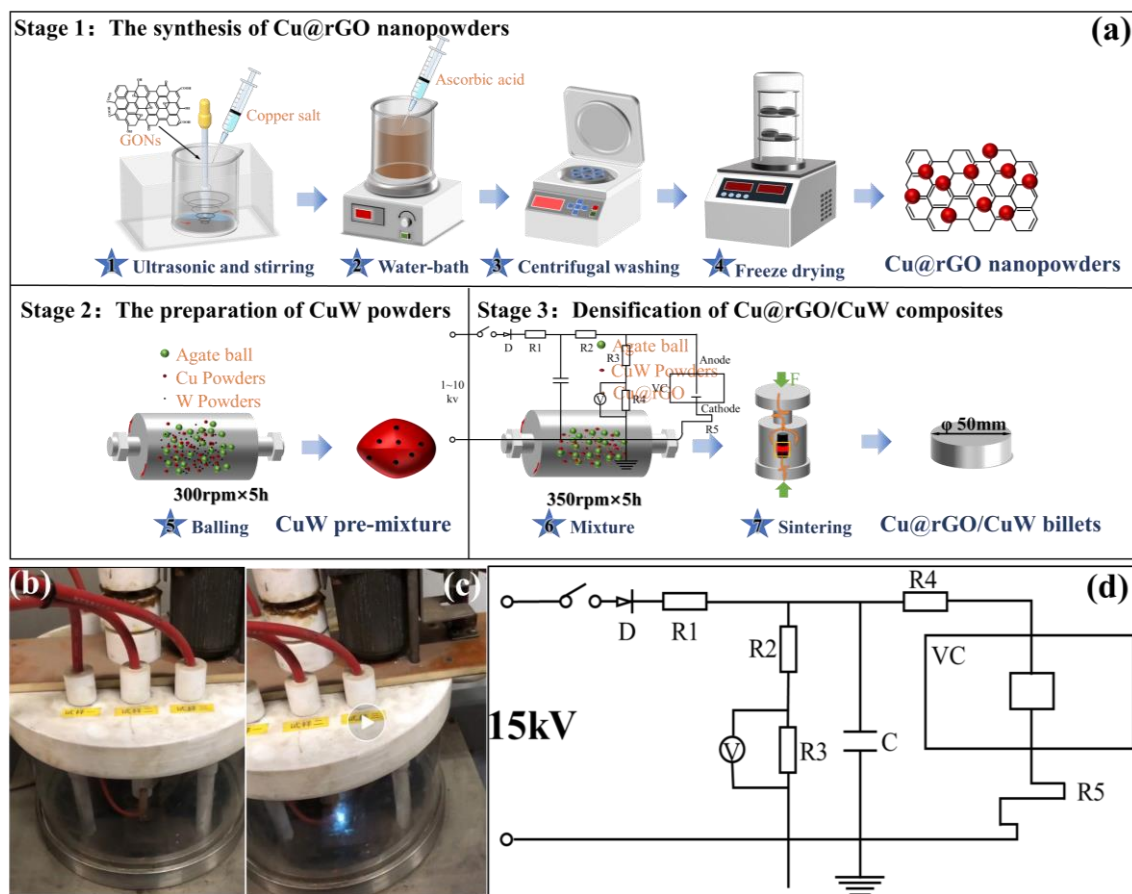
1 with Cu nanoparticles (Cu@rGO) using an one-step chemical co-reduction route. All
2 the chemicals were used directly without further purification. **Figure 2a** shows the
3 detailed schematic illustrations of the fabrication process of Cu@rGO nanoparticles
4 formed by one-step chemical co-reduction route. Macrophotograph of the fabricated
5 Cu@rGO nanopowders in **Figure 1(c)** shows the color was changed from brown (GO's
6 color) into black after one-step chemical co-reduction method. Scanning electron
7 microscope (SEM) image of as-synthesized Cu@rGO nanopowders in **Figure 1(d)**
8 shows that the Cu nanoparticles (confirmed by EDS analysis in **Figure 1e** and **Figure**
9 **1f**) with a spherical one or a hexahedron one are uniformly distributed on the surfaces
10 of wrinkled rGO nanosheets.



11 **Figure 1(a)** TEM image of graphene oxides, inset showing the high resolution TEM of
12

1 GO; (b) Selected electron diffraction pattern; (c) Macrophotograph of fabricated
 2 Cu@rGO nanopowders; (d) SEM image of Cu@rGO nanocomposites, (e) EDS point
 3 analysis of marked region in Figure (d), (f) EDS mapping results of Figure (d),
 4 respectively.

5



6

7 **Figure 2(a).** Schematic illustration of microstructure of Cu@rGO/CuW composites
 8 fabricated using combination of *in-situ* co-reduction method, ball milling and SPS; (b)
 9 the state before the breakdown; (c) the momentary state of the breakdown and (d)
 10 experimental circuitry diagram of vacuum struck, respectively.

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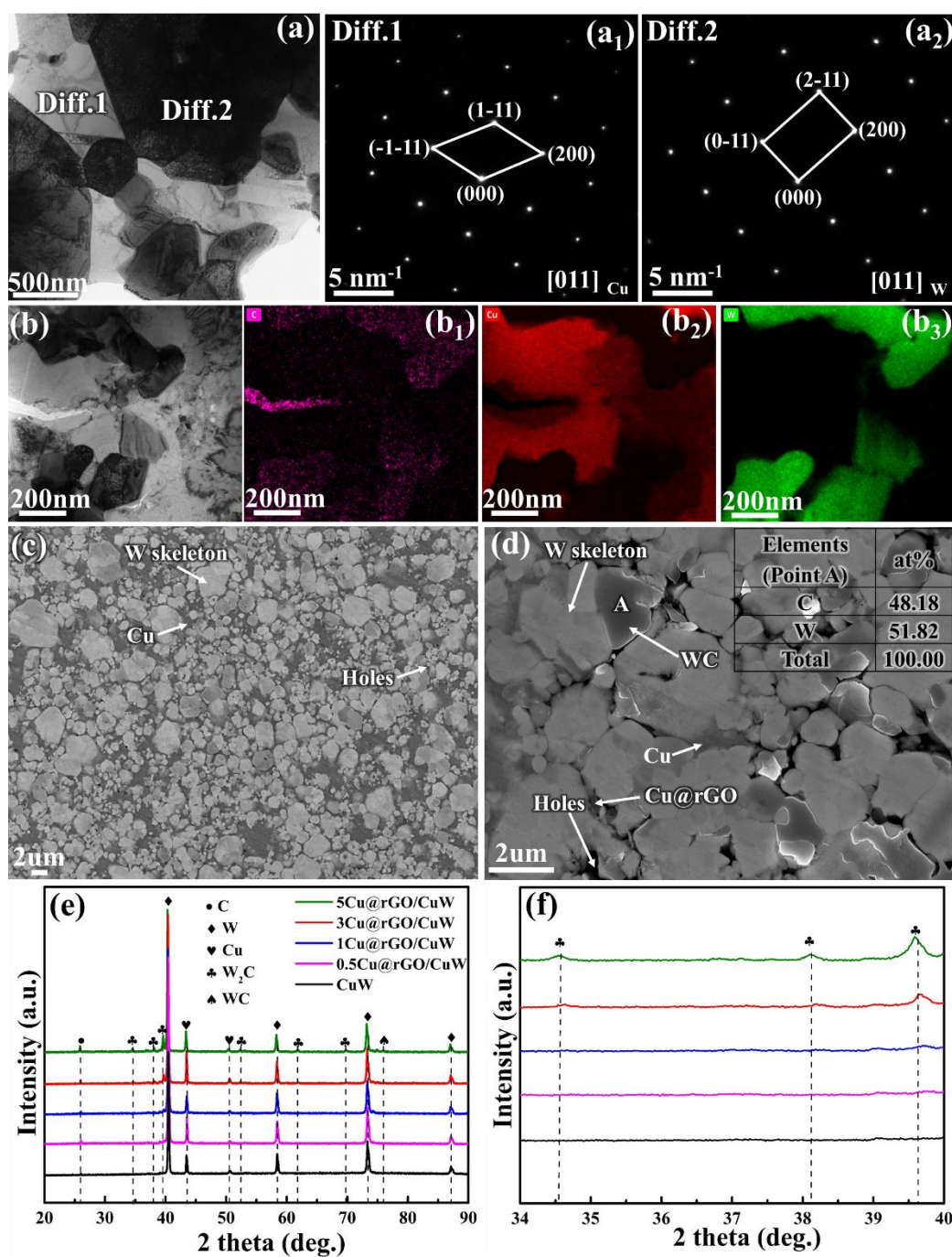
12 Commercially available tungsten powders (1~2 μm in size, purity of 99.9%) and

1 electrolytic copper powder (4~7 μm in size) were selected to fabricate CuW composites.
2 The copper modified reduced graphene oxides/CuW composites (i.e. Cu@rGO/CuW)
3 were fabricated using combined *in-situ* co-reduction method [23, 24], ball milling and
4 spark plasma sintering (SPS) technology, as shown in **Figure 2**. Mixed powders with
5 30 wt% copper and 70 wt% tungsten were used as the initial sources, which were firstly
6 blended in a drum ball mill (QM-5) for 5 hrs at a rotating speed of 300 rpm, with a mass
7 ratio of ball and powder of 3:1. During this process, the soft copper powder particles
8 were deformed and the W particles were adhered onto the surfaces of copper particles.
9 Subsequently, the milled CuW mixtures with various Cu@rGO nanopowders were
10 further milled in a drum ball milling machine at 350 rpm for 5 hrs with a ball to powder
11 ratio of 1.5:1. Finally, the Cu@rGO/CuW composite powders were densified using SPS
12 at 1000 $^{\circ}\text{C}$ for 10 min with an axial pressure of 45 MPa in a vacuum atmosphere. In
13 addition, CuW composites as a control group were fabricated by the same process.

14 For more detailed microstructural information of the constituting phases in the
15 sintered composites, the 3wt%Cu@rGO/CuW composites was taken as an example for
16 further TEM analysis as shown in **Figure 3**. According to the bright field image shown
17 in **Figure 3a**, no visible gaps or cracks were found in the composite, revealing that a
18 good bonding formed between W and Cu phases. The corresponding selected area
19 diffraction (SAED) patterns shown in **Figure 3a₁** and **Figure 3(a₂)** confirm the face
20 centred cubic phase of Cu and body-centred cubic phase of W. However, rGOs with
21 a width of ~ 100 nm are well embedded and bonded with the Cu matrix without the
22 presence of cracks, as shown in **Figure 3b**. EDS mapping analysis reveals the diffusion

1 reaction between rGO and W matrix during the sintering, as shown in **Figures 3b₁** and
2 **3b₃**. The formation of tungsten carbides in CuW composites is in favor of improvements
3 of their mechanical and arc erosion resistance [14, 25, 26]. As shown in **Figure 3c**, the
4 microstructure of the CuW composites without adding Cu@rGO is relatively uniform
5 and dense. The gray quasi-spherical phase comprises tungsten particles, which form
6 into a continuous skeleton structure. The black network structure is the copper phase.
7 However, when adding Cu@rGO nanopowders, as shown in **Figure 3d**, the quasi-
8 spherical tungsten particles are changed into angular ones, indicating that the addition
9 of Cu@rGO nanopowders resulted in new tungsten phases. EDS analysis in **Figure 3d**
10 and XRD results in **Figures 3e** and **3f** further confirm the formation of tungsten carbide
11 particles owing to their lower Gibbs free energy.

12



1
2 **Figure 3.** (a) Bright field image of CuW composites; (a₁) and (a₂) SEAD pattern taken
3 from marked region Diff. 1 and Diff. 2 in Figure a; (b) Bright field image of
4 3wt%Cu@rGO/CuW composites and (b₁) ~ (b₃) corresponding EDS mapping results;
5 SEM images of (c) CuW composites and (d) 3wt% CuW composites; (e) X-ray
6 diffraction analysis results of CuW composites with Cu@rGO nanopowders; (f)
7 enlarged XRD patterns at $2\theta = 34^\circ \sim 40^\circ$ in Figure (e), respectively.

1 2.2 Characterizations

2 Microstructures of Cu@rGO and Cu@rGO/CuW composites were characterized
3 using a field emission scanning electron microscope (FESEM, Zeiss GeminiSEM 500)
4 and a high-resolution transmission electron microscope (HRTEM, JEOL JEM-
5 2100Plus) with selected area electron diffraction (SAED). For TEM sample preparation,
6 about 0.2 mm flake was obtained using a wire-cutting machine, mechanically ground
7 to a thickness of 30~50 μm using a metallographic sand paper, and was then carried out
8 using a Gatan PIPS 691 ion milling system with a time of 6 hrs and argon ions were
9 used in the ion etching with the accelerating voltage in the range of 2.5–5 keV and with
10 an incident angle in the range of 4–10°. The phase compositions were investigated using
11 X-ray diffraction (XRD) with a Cu $K\alpha$ radiation.

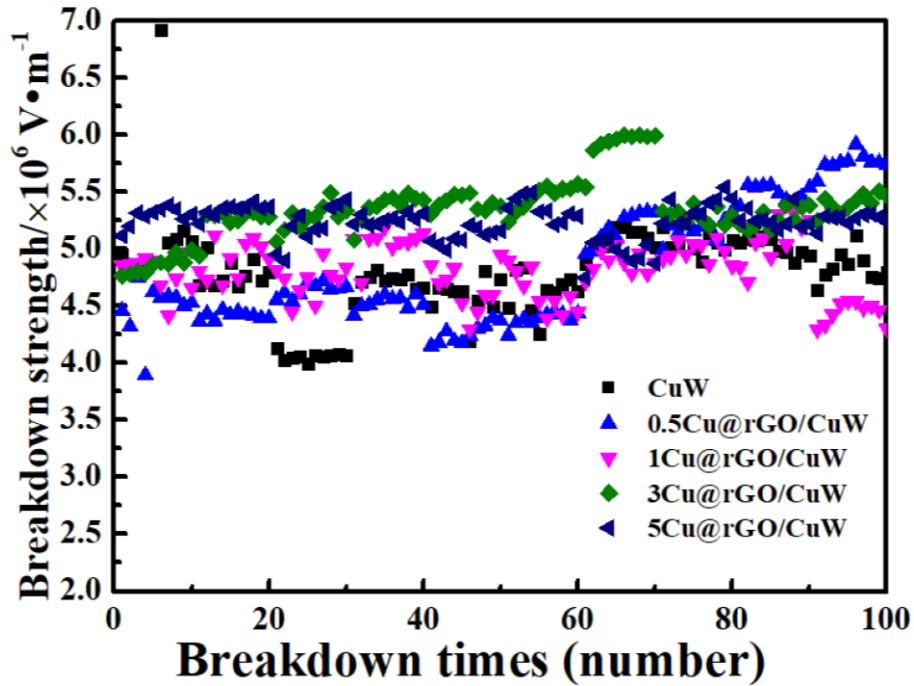
12 Schematic illustrations of the testing facility and experimental circuitry diagram
13 of vacuum struck are shown in **Figures 2b ~ 2d**. A pure W needle with a tip radius of
14 3 mm was used as an anode. The composite samples were first polished mechanically,
15 and then placed in the vacuum chamber as a cathode. SF₆ gas was used as an arc
16 extinguishing medium when the electrical contact was switched on and off with a
17 voltage of 15 kV. Subsequently, the anode electrode was moved to the surface of the
18 sample at the speed of 0.01 mm/min until the electric breakdown occurred between
19 electrodes. As shown in **Figure 2c**, a strong bright electrical arc was observed at the
20 surface of samples, resulting in the complex physical and chemical reactions during the
21 arc breakdown process. The distance for migration between cathode and anode was
22 measured using a digital micrometer. After the arc extinguished, the electrical

1 breakdown test was repeated 100 times. The breakdown strength was obtained using
2 the breakdown voltage divided by the breakdown gap [27]. The surface morphologies
3 of samples after arc breakdown were determined using the SEM with an energy
4 dispersion spectrometer (EDS). X-ray photoelectron spectroscopy (XPS, Thermo
5 Fisher ESCALAB Xi+) was used to examine the chemical states of surface elements
6 after arc breakdown. The 3D morphology of the ablated surface after 100 times arc
7 breakdown was characterized using a three-dimensional profilometer (America, KLA
8 Tencor-MicroXAM-800).

9

10 **3. Results and discussion**

11 Arc erosion resistance is a key parameter for CuW composites to be used as the
12 electrical contact materials [27]. Relationships between arc breakdown strength and
13 number of break-down of the specimens with various Cu@rGO concentrations were
14 obtained under a voltage of 15 kV. The obtained results shown in **Figure 4** reveal that
15 the dielectric breakdown strength of CuW composites was relatively stable over 100
16 electrical breakdown cycles, and the average strength was $\sim 4.5 \times 10^6$ V/m. However,
17 with an increase of the Cu@rGO contents, the average dielectric breakdown strength
18 of the specimens was increased. The calculated average dielectric breakdown strengths
19 of the specimens with 0.5 wt%, 1.0 wt%, 3 wt% and 5.0 wt% Cu@rGO were 4.6×10^6
20 V/m, 4.8×10^6 V/m, 5.8×10^6 V/m, and 5.5×10^6 V/m, respectively, revealing that the
21 addition of Cu@rGO nanopowders has enhanced the average dielectric strength of
22 CuW composites, which is 28.9% larger than that of CuW matrix.



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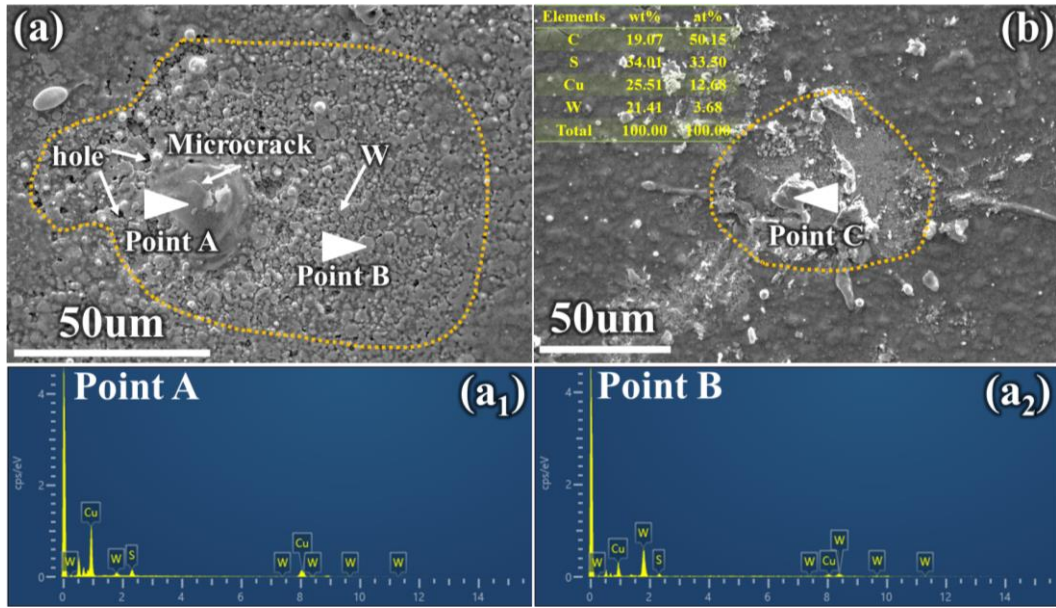
2 **Figure 4.** The relationships between electrical breakdown strength and breakdown
 3 times of the CuW composites without and with various Cu@rGO nanopowders,
 4 respectively.

5

6 In order to determine the first breakdown phases and places of arcs, surface
 7 morphologies of CuW alloy and Cu@rGO/CuW composites after the first arc
 8 breakdown in SF₆ atmosphere were observed using SEM, and the obtained image is
 9 shown in **Figure 5**. A circular arc erosion region with 0.0176625 mm² can be observed
 10 on the surface of CuW alloy. A micro-size hole was observed in the arc erosion region
 11 owing to the evaporation of the low melting point Cu phase. The EDS of irregular
 12 protrusion (denoted by point A in **Figure 5a**) in **Figure 5(a₁)** reveals the main peaks of
 13 Cu and a small quantity of W owing to the condensation of copper vapor. Whereas, the
 14 arc erosion craters of Cu@rGO/CuW composites with 0.005024 mm² are smaller and
 15 shallower than those of CuW composites, as compared **Figures 5a** and **5b**. It can be

1 seen in **Figure 5b** that the breakdown pits on the surface of Cu@rGO/CuW composites
2 are dispersed uniformly, and the protrusion is smaller than that in **Figure 5a**. EDS result
3 of point C (Inset in **Figure 5b**) reveals a dominant carbon main peaks with a small
4 quantity of Cu and W elements, revealing that the graphene was subjected arc
5 breakdown during the first breakdown.

6 According to the field emission theory [28, 29], under the action of external
7 electric field, interface atoms and electrons are easy to escape due to the weaker binding
8 force which comes from the CuW bonding interface. Besides, the occurrence of
9 breakdown is closely related to the work functions of metal elements under the same
10 electrical breakdown conditions, and Cu phase has a lower work function. Therefore,
11 discontinues electrical breakdowns of CuW composites usually occurs on the Cu/W
12 phase interfaces and copper-rich regions, and the arc cannot be moved rapidly on the
13 surface of CuW alloy as shown in **Figure 5a**. Whereas, when the Cu@rGO
14 nanopowders was introduced into CuW matrix, the interfaces of Cu/W phase were
15 strengthened owing to the diffusion of carbon atom and formation of tungsten carbide
16 particles in the composites, which makes the Cu/W interface forming a strong
17 metallurgical bonding. As a result, the Cu/W interfacial energy is decreased because the
18 element of C is dissolved and diffused into W phase at the interface [30].



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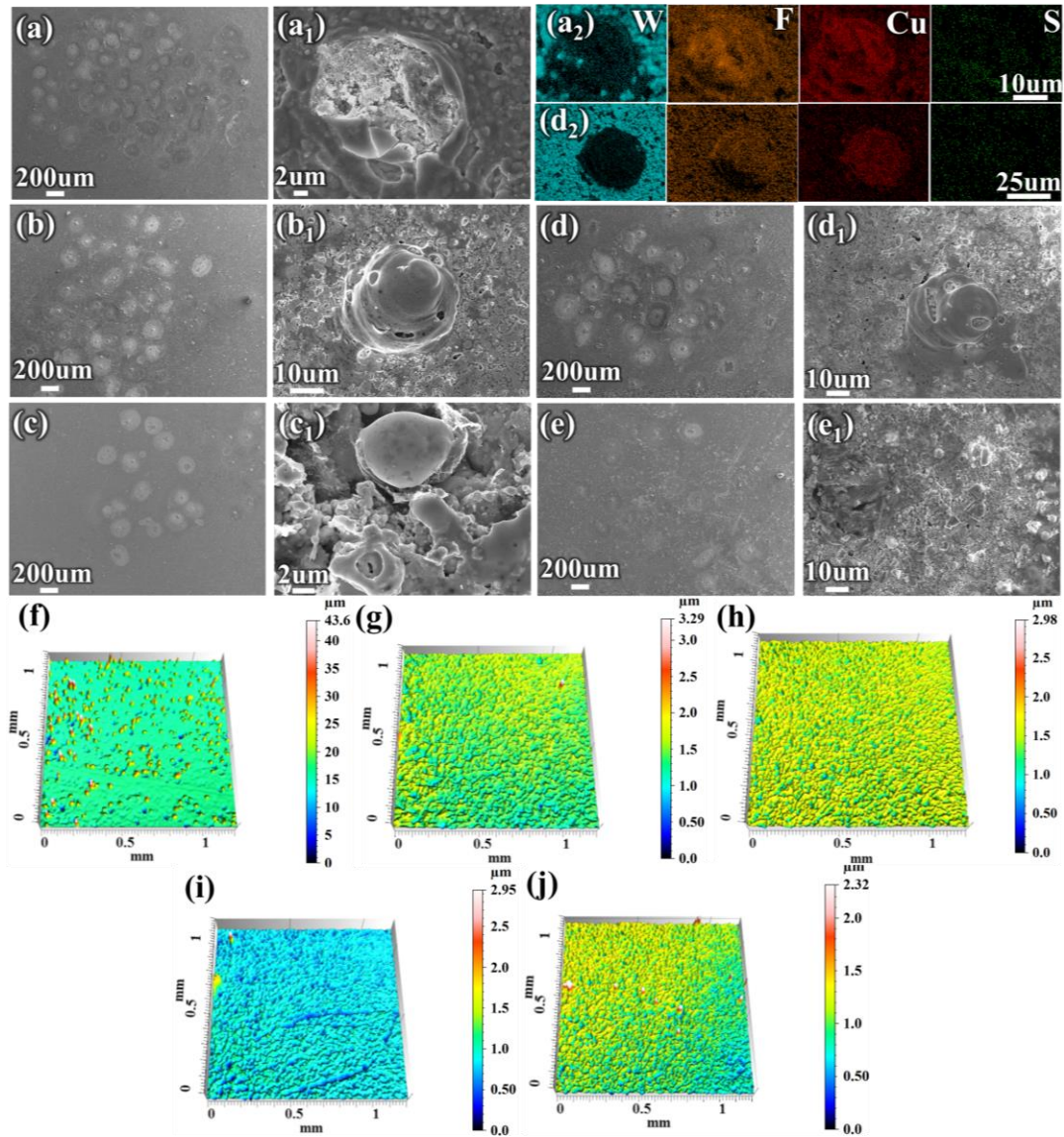
2 **Figure 5.** SEM images of CuW and Cu@rGO/CuW composites after the first electrical
 3 breakdown (a) CuW composites, (a₁) and (a₂) EDS analysis of point A and point B
 4 marked in Figure (a), (b) Cu@rGO/CuW composites, inset showing the EDS result of
 5 point C in Figure (b), respectively.

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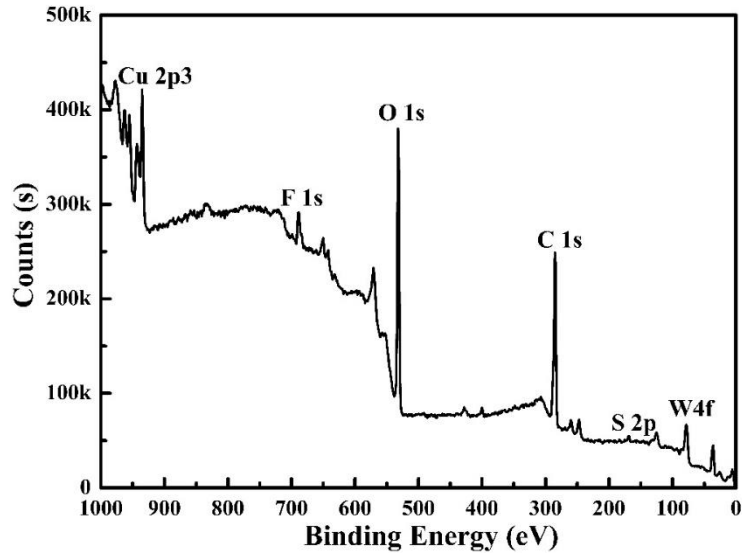
7 SEM micrographs of sample surfaces with different contents of Cu@rGO
 8 nanopowders after 100 breakdown cycles are presented in **Figure 6**. It can be seen from
 9 **Figures 6(a)~(a₁)** that the most serious arc erosion with rough morphology occurred in
 10 the central region just underneath the tungsten anode tip. EDS mapping analysis in
 11 **Figure 6(a₂)** shows that the droplets and their surrounding areas have mainly the
 12 elements of F, S, Cu and W. The F and S elements are from the protective atmosphere
 13 of SF₆ in the working environment of the electrical contact, indicating SF₆ extinguish
 14 medium has been involved in surface chemical reactions under the high temperature
 15 arc process [23], which eventually causes the CuW contacts to lose their functions of
 16 conducting in the course of long-term service. Whereas, with an increase in the

1 Cu@rGO contents, the breakdown pits were decreased in depth and became
2 increasingly diffused, as compared with **Figures 6(b) ~ (e)**. The above results indicate
3 that the ability of the specimens to disperse arc became dramatically enhanced owing
4 to the presence of Cu@rGO.

5 XPS analysis of the Cu@rGO/CuW composites after breakdown for 100 cycles in
6 **Figure 7** shows the presences of the C1s and Cu 2p³ peak, further revealing the
7 graphene was reserved after the electrical breakdown. Moreover, the 3D profiles
8 analysis in **Figure 6(f~j)** reveals that addition of Cu@rGO nanopowders can
9 significantly reduce the roughness of samples after arc ablation. This is mainly due to
10 the following reasons. (a) The ceramic particles are generated by the reaction between
11 the tungsten skeleton and rGO during the high temperature sintering process (**Figures**
12 **3d ~ f**); (2) The superior thermal conductivity of graphene improves the wetting
13 capabilities of matrix during the arc breakdowns process, (3) Cu@rGO nanophases
14 have a lower working function compared with those of W and Cu phases [20], and (4)
15 There is an enhanced W/Cu interfacial bonding owing to diffusion and *in-situ* reactions
16 (**Figure 3b**).



1
2 **Figure 6.** SEM image of surface morphologies and corresponding 3D profiles after
3 electrical breakdown for 100 cycles of CuW composites with different contents of
4 Cu@rGO (a, f) 0 wt%, (b, g) 0.5 wt%, (c, h) 1 wt%, (d, i) 3wt% and (e, j) 5wt%,
5 respectively.



1

2 **Figure 7.** XPS survey spectrum of the Cu@rGO/CuW composites after electrical
3 breakdown for 100 cycles.

4

5 For the CuW system, the melting point of Cu is much lower than that of W,
6 therefore, the Cu phases around the W skeleton structures are melt and evaporated to
7 the contact surface during the arc-erosion process as shown in **Figure 5a**. With the
8 increased breakdown time, the W skeleton structure is constantly peeled off (**Figure**
9 **8b4**). After arc extinguished, obvious ablation pits are formed on the surface of CuW
10 composites (**Figure 8a6**). Therefore, as reported in references [5, 29, 31, 32], the main
11 mechanisms of ablation failure for the CuW contact materials are splashing and
12 evaporation of the Cu phases, and ablation of the W skeleton (**Figure 5a** and **Figure**
13 **8a**).

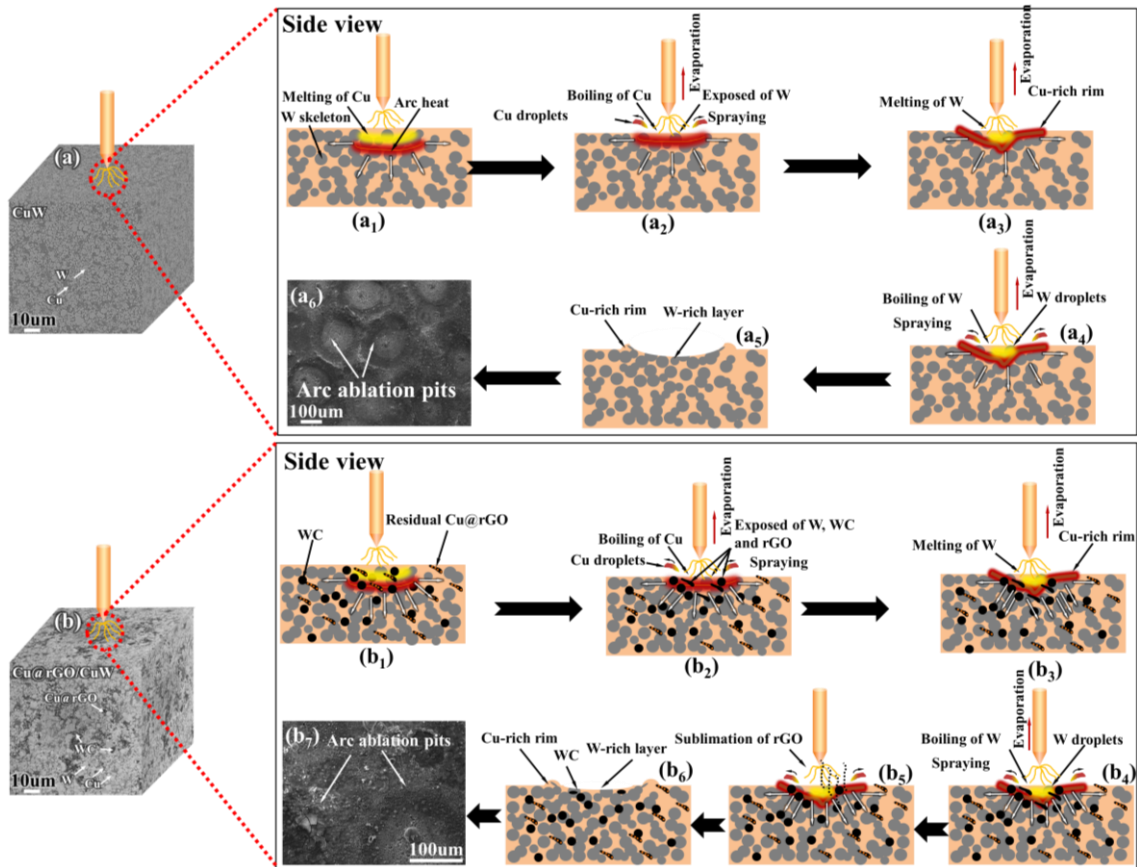
14 There are many factors that may affect the arc ablation properties of electrical
15 contact materials, such as the roughness of contact surface, the shape and size of the
16 contact materials and the external magnetic field [23]. However, the most important

1 ones are determined by the physical properties of the contact material itself, such as its
2 work function, electron potential barrier and microstructure characteristics. For the
3 Cu@rGO/CuW composites as shown in **Figures 8b ~ b7**, when the Cu molten pool is
4 firstly produced under the effect of high arc energy, Cu@rGO will increase the viscosity
5 of the molten metal Cu owing to Cu@rGO's low density, so it will float on the surface
6 of molten metal Cu (**Figure 8b₁ ~ b₂**). This can stabilize the molten pool, thus inhibiting
7 the flow and splashing of molten metal to reduce arc ablation.

8 Furthrmore, Cu nanoparticles are not completely coated on the surface of reduced
9 graphene oxides nanosheets (**Figure 1d**), the C atoms (derived from rGO) can be *in-*
10 *situ* formed from W skeleton during the SPS owing to low Gibbs free energy of tungsten
11 carbide. According to the work function theory, the arc would occur firstly on the phase
12 with a smaller work function during dielectric breakdown process. The work functions
13 of tungsten, copper, graphene and tungsten carbide are 4.54 eV [14], 4.36 eV [14], 4.2
14 eV [33] and 3.79 eV [14], respectively. Therefore, breakdowns may preferentially occur
15 on the tungsten carbide phase and Cu@rGO phase instead of Cu phase, thus slowing
16 down the spatter of copper. When the arc breakdown occurs, graphene cladding can
17 restrain the electron emission, causing the efficeint transfer of heat generated by the arc.

18 Besides, under the heat effect of arc, the tungsten carbides with a good stability at
19 a high temperature are distributed in the CuW matrix, and prevent the flow or the
20 accumulation of molten copper, and also avoid the large area splash of copper liquid.
21 On the other hand, the tungsten carbide ceramic particles with high melting points can
22 be used as the cores of heterogeneous crystal nucleation, which shortens the

1 solidification time of molten copper liquid and reduces the size of particles formed on
 2 the composite surface (**Figure 8b₆**). Both the above-mentioned factors would contribute
 3 to the improved anti-ablation properties and extended service life of Cu@rGO/CuW
 4 composites (**Figure 8b₇**).



5
 6 **Figure 8.** Schematics illustration of the ablation mechanisms of (a~a₆) CuW composites
 7 [29] and (b~b₇) Cu@rGO/CuW composites, respectively.

8
 9 **4. Conclusions**

10 In summary, a one-step and co-reduction chemical process strategy was developed
 11 to directly modify reduced graphene oxides with Cu nanoparticles for preparing high-
 12 performance CuW composites. Tungsten carbide phases were formed through the *in-*
 13 *situ* reaction of rGO and tungsten during the SPS process. The CuW composites with

1 Cu@rGO showed significantly improved arc erosion resistance. Compared with CuW
2 composites, the ablation area of Cu@rGO/CuW composites was smaller and the
3 ablation craters were shallower, and the average dielectric vacuum breakdown strength
4 of the CuW composites with 3 wt% Cu@rGO was increased by 28.9% in comparison
5 with the CuW composites without Cu@rGO. The arc breakdown mechanism of
6 Cu@rGO/CuW composites are summarized as following. (1) Addition of Cu@rGO
7 increases the viscosity of molten metal Cu, inhibiting the flow and splashing of molten
8 metal; (2) Lower work functions of carbon (i.e. rGO) and tungsten carbide restrains the
9 electron emission; and (3) Tungsten carbides with their good stability and high melting
10 point shorten the solidification time of molten copper liquid, thus extending service life
11 time of Cu@rGO/CuW composites.

12

13 **Declaration of Competing Interest**

14 The authors declare that they have no known competing financial interests or
15 personal relationships that could have appeared to influence the work reported in this
16 paper.

17

18 **Acknowledgments**

19 The authors would like to acknowledge the financial support from National
20 Natural Science Foundation of China (No. 51901192), the Shaanxi Science Foundation
21 For Distinguished Young Scholars (2020JC-50), Key Research and Development
22 Projects of Shaanxi Province (No. 2019GY-164), Foundation of Northwest Institute for

1 Non-ferrous Metal Research (No. YK2020-9), as well as International Exchange Grant
2 (IEC/NSFC/201078) through Royal Society and National Science Foundation of China
3 (NSFC). We thank Master Hang Xue and Ning Tian at Xi'an Rare Metal Materials
4 Institute Co., Ltd for assistance in carrying out arc ablation experiment, and Professor
5 Wenge Chen at Xi'an University of Technology for the usage of arc ablation equipment.

6

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