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# Patterning of the Surface Electrical Potential on Chalcogenide Glasses by a Thermoelectrical Imprinting Process

Ricardo Alvarado, Lara Karam, Redouane Dahmani, Antoine Lopicard, Florian Calzavara, Andréa Piarristeguy, Annie Pradel, Thierry Cardinal, Frédéric Adamietz, Evelyne Fargin, et al.

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7 Patterning of the surface electrical potential on  
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11 chalcogenide glasses by a thermo-electrical  
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15 imprinting process  
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3 KEYWORDS:  
45 Surface potential, Chalcogenide glasses, poling, glass structure  
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1516 ABSTRACT  
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1920 The development of novel sensing systems requires breakthroughs in the conception of  
21 multifunctional materials. In this sense, while extensive research has been dedicated to the  
22 individual tuning of the electrical or optical properties of different materials, the combination of  
23 both features would result in a promising field of research that would further extend opportunities  
24 for engineering novel function in sensor geometries. In the present work, we employed a highly  
25 attractive optical material for mid-infrared (MIR) sensing (chalcogenide glasses, ChG) and  
26 focused on the spatial control of its surface electrical potential via a thermoelectrical imprinting  
27 process. Different glass compositions based on the system Ge-Sb-S-Na were prepared by varying  
28 the sulfur stoichiometry and the sodium content. Each glass was thermally poled using electrodes  
29 with specific patterns, and subsequent structural modifications and surface electrical potential were  
30 then evaluated via Raman spectroscopy and Kelvin Probe Force Microscopy (KPFM). Raman  
31 cartographies show structural modifications attributed to alkali depletion following the patterns of  
32 the electrodes used for the imprinting process. Furthermore, KPFM measurements show clearly  
33 defined motifs on the electrical potential which are associated to charges implanted into the glass  
34 matrix. It was shown that the surface potential can vary in sign within an amplitude range of 10V  
35 and exhibit patterning at the micrometer scale. We observed that the efficiency of the surface  
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3 potential imprinted was dramatically impacted by the glass' sulfur and sodium content. Our results  
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5 demonstrate for the first time the use of a one-step process, thermal poling, for large scale  
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7 patterning of the surface potential of ChG creating a multi-functional material.  
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## 10 11 12 Introduction

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15 In the context of lab-on-chip and other nanotechnological systems, control of a  
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17 material's electrical functionality is a valuable lever that, if tunable, offers design flexibility  
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19 to enhance device performance. The patterning of the electrical potential on the surface  
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21 at micrometer or nanometer scale has been proven to be an effective strategy to  
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23 assemble nanoparticles or nano-objects precisely at a desired location on a substrate  
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25 through electrophoresis<sup>1</sup>. Xerography is a well-recognized technique that uses  
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27 electrostatic forces to allow selective attachment of toner particles<sup>2</sup>. Since its early  
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29 development, the technique has evolved considerably such that one can now talk about  
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31 micro<sup>3</sup> or even nanoxerography<sup>4-6</sup>. The electrical patterning of the substrate, an  
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33 appropriate charge carrier or electret, can be realized through atomic force  
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35 microscopy<sup>1,4,6</sup>, electron beam<sup>7</sup> or focused ion beam<sup>8</sup> with a lateral resolution of about a  
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37 hundred nanometers. The main limitation of these techniques is that they are time  
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3 consuming when large areas (centimeter square) are considered. To circumvent this  
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7 problem parallel printing processes to pattern charges on the electret over sizes  
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10 approaching a centimeter square, during a single process cycle have been  
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13 developed<sup>3,5,9,10</sup>. This stamp-like process doesn't deteriorate the resolution and even  
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16 lower resolution (to 50 nm) can be obtained when a localized fringing field<sup>10</sup> is induced to  
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19 direct the nanoparticle assembly. Guiding the growth of nano-objects on specific sites is  
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22 not the only possibility that offers surface potential patterning. Rapid and reversible  
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25 changes induced by potential to alter the focal length of liquid lenses has been achieved  
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28 by tuning the electrochemical desorption of self-assembled monolayers (SAMs) onto a  
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31 metal surface<sup>11</sup>. It is also possible to induce dynamic changes in interfacial properties  
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34 such as wettability by generating conformational transitions of single layered amphiphilic  
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37 molecules deposited onto a metallic substrate: a change in the sign of the potential leads  
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40 to a transition between hydrophilic (molecules are straight) and hydrophobic (molecules  
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43 are bent) nature of the surface<sup>12</sup>.  
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52 Research efforts towards the development of components merging multiple  
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55 functionalities are more and more essential. In this context, the combination of electrical  
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3 and optical functionalities on a single material represents a great benefit. Optical glasses  
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7 such as chalcogenides (ChG) are particularly interesting for sensing applications, as they  
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10 are transparent in the MIR, the spectral region in which the organic molecules fingerprint  
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13 can be found. In this study, we demonstrate for the first time the patterning of the surface  
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16 electrical potential of Ge-Sb-S-Na glasses via a one-step imprinting process covering  
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19 large areas at once (centimeter square scale): thermal poling. This process consists in  
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22 applying a DC field to a glass heated at moderate temperature; the glass is brought back  
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25 to ambient temperature with the DC field still on. This freezes the electrical constraint at  
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28 the surface of the glass and induces several other modifications: depletion of cations<sup>13</sup>,  
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31 structural rearrangements<sup>14–16</sup>, modification of the surface reactivity<sup>17–19</sup>, wettability  
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34 change<sup>20</sup>, surface potential<sup>21</sup>, changes of the optical properties (modification of the  
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37 index<sup>22</sup> and induced second order nonlinear response<sup>23–27</sup>) as well as the formation and  
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40 patterning of glass-metal nanocomposites<sup>28–31</sup>. The patterning of these changes has been  
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43 realized using structured electrodes that act like stamps to control the glass' surface  
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46 properties leading to the design of functional elements such as linear<sup>19</sup> and nonlinear  
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49 diffraction gratings<sup>32</sup> or arrays of micro-lenses<sup>33</sup>. In this study, we present evidence of the  
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3 viability in patterning of surface electrical potentials in glasses at the micrometer scale.  
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7 Through parallel studies illustrating the influence of the glass' sodium and sulfur content  
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10 on these surface properties combined with Raman microscopy, we proposed a  
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13 mechanistic explanation as to the origin of the surface potential in order to generate  
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17 effective and stable patterns for future applications.  
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## 20 **Experimental Methods**

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23 *Synthesis of ChG glasses:* ChG glasses were prepared using high purity elemental Ge, Sb and S  
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25 (Alfa Aesar, 99.999%). Anhydrous sodium sulfide (purity unspecified) was used to incorporate  
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27 sodium as doping agent. Based on the stoichiometric composition  $\text{Ge}_{25}\text{Sb}_{10}\text{S}_{65}$ , different samples  
28  
29 were prepared by varying either the S to Ge ratio (i.e. increasing the sulfur content) or by keeping  
30  
31 the ratio between the other components constant and adding sodium. The nominal compositions  
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33 of the different glasses are reported in table 1 as well as the name by which they will be referred  
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35 to in the rest of the present work. Two series of glasses were synthesized: (i) A, B, C, D with  
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37 constant sodium content to evidence the influence of sulfur and (ii)  $\text{Na}_0$ ,  $\text{Na}_2$ ,  $\text{Na}_{5.5}$  with constant  
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39 S to Ge ratio to discuss the influence of the sodium content. Raw materials were weighed out in a  
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41 glove box under a nitrogen atmosphere before being inserted in a quartz ampule that was, then,  
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43 put under vacuum ( $10^{-2}$  mbar) and sealed using an oxygen-methane torch. A rocking furnace was  
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45 employed to ensure the homogeneity of the mixture. The temperature was increased at a rate of  
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48  $1\text{ }^\circ\text{C min}^{-1}$  up to  $850\text{ }^\circ\text{C}$  and left at this value for 12 h for homogenization. After this period of  
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51 time, the rocking of the furnace was stopped and the temperature was set to gradually decrease to  
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54  $750\text{ }^\circ\text{C}$  (quenching temperature). The sample was then quenched in water and annealed for 6 h,  
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10 °C under its glass transition temperature ( $T_g$ ); the  $T_g$  of the different glasses ranging between 250 and 350 °C. Finally, glasses were cut and polished into ~1 mm thick disks of 1 cm of diameter.

	Sample's name	Composition (at. %)				S/Ge ratio
		Ge	Sb	S	Na	
Influence of the S/Ge or sulfur content	A	24.5	10	64.5	<b>1</b>	<b>2.6</b>
	B	22	10	67	<b>1</b>	<b>3</b>
	C	20	10	69	<b>1</b>	<b>3.4</b>
	D	17	10	72	<b>1</b>	<b>4.2</b>
Influence of the sodium content	Na <sub>0</sub>	22.5	10	67.5	<b>0</b>	<b>3</b>
	Na <sub>2</sub>	22	9.8	66.2	<b>2</b>	<b>3</b>
	Na <sub>5.5</sub>	21.3	9.4	64.8	<b>5.5</b>	<b>3</b>

Table 1: Atomic composition of the different glasses of interest and the name they are referred to.

*Vibrational spectroscopy:* Raman spectra were acquired using an XploRA PLUS (Horiba) spectrometer with a laser source at 785 nm focused on the sample via a 50x objective lens. Cartographies were acquired using a 100x (N.A. 0.9) objective lens on an area of 80 x 60  $\mu\text{m}$  with a lateral resolution of ~2  $\mu\text{m}$ .

*Thermoelectrical imprinting:* Electrical patterning was performed under a nitrogen atmosphere by pressing the ChG glass sample between two electrodes: a structured electrode used as a stamp located at the anode, and a silicon electrode at the cathode. A coverslip was used as a sacrificial layer in between the glass sample and the cathode. The structured electrode consists of a 100 nm thick Ti-Pt metallic films deposited on a silicate glass slide. The patterned electrode was prepared by a classical lithography process resulting in squared motifs (40x40 or 30x30  $\mu\text{m}$ ) leaving a 4  $\mu\text{m}$  width electrically conductive Pt grid. The system was then heated to 210 °C, after reaching thermal stabilization, an electric field was applied at 375  $\text{V}\cdot\text{min}^{-1}$  up to 0.9 kV. The temperature and DC

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3 bias were kept constant for 30 min, then, the system was cooled down to room temperature before  
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5 removal of the electric field.  
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8 *Characterization of surface electrical potential:* Measurements were performed at 100 nm from  
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10 the surface by an AFM Dimensions Icon (Bruker) on PeakForce KPFM mode using a PFQNE-AL  
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12 cantilever, over a 55 x 55  $\mu\text{m}$  area and a resolution of 1024 points per line.  
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## 14 **Results**

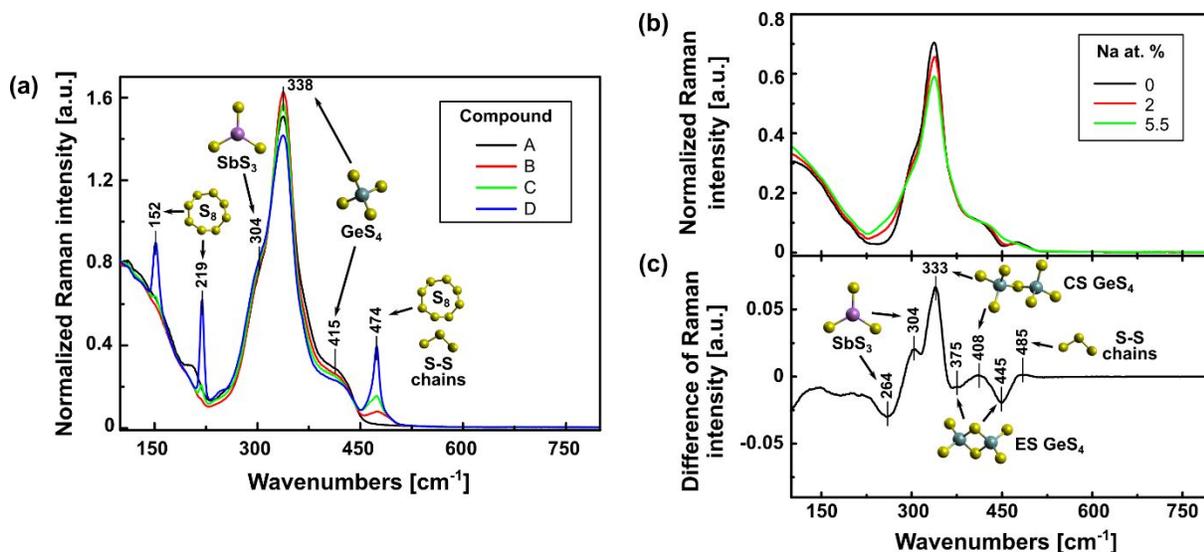
### 15 *Structural analysis of pristine glasses*

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17 Raman spectra of each pristine sample were acquired and normalized over their total integrated  
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19 area (**Figure 1**). Different models to describe  $\text{GeS}_2$  glass networks have been proposed in the  
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21 literature<sup>34–37</sup> resulting in different assignments for the characteristic Raman bands associated with  
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23 this network.<sup>38–40</sup> A recent article regarding multimodal structural characterization of Ge-S-I<sup>39</sup> as  
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25 well as previously reported DFT calculations<sup>40</sup> were used for the assignments in this work. The  
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27 spectra presented in **Figure 1a** show different contributions associated with the stretching modes  
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29 of Ge-S bonds in corner sharing (CS)  $\text{GeS}_4$  tetrahedra<sup>39,40</sup> around  $340\text{ cm}^{-1}$  and  $415\text{ cm}^{-1}$  as well  
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31 as Sb-S stretching modes in  $\text{SbS}_3$  pyramids<sup>41</sup> resulting in a small shoulder around  $300\text{ cm}^{-1}$ . The  
32  
33 most striking variation when the amount of sulfur increases (i.e. going from glass A to D) is the  
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35 apparition of new bands at  $152\text{ cm}^{-1}$ ,  $219\text{ cm}^{-1}$  and  $474\text{ cm}^{-1}$ , they correspond to S-S bonds.<sup>39,42,43</sup>  
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37 The first two bands, present only when the sulfur content exceeds 70 at.%, are linked to the  
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39 formation of sulfur rings ( $\text{S}_8$ ) inside the glass matrix.<sup>39,42</sup> Another evolution when increasing the  
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41 sulfur content, is the regular decrease of the contribution of the Ge related entities ( $340$  and  $415\text{ cm}^{-1}$ ).  
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43 This is consistent with the decrease of the Ge content in the composition when going from  
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45 composition A to D. The Raman intensity of the modes related to Sb are not affected which is  
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47 consistent with the fact the Sb content stays constant. More interestingly, the modes related to  
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3 either the Ge or Sb entities are observed at the same wavenumber, no frequency shift can be  
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5 observed when the sulfur content increases. These observations suggest that the surplus of sulfur  
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7 does not alter the glass forming units but rather the interconnections among them by establishing  
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9 new homopolar S-S linkages. This results in the creation of S-S chains up to a given sulfur content  
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11 (close to 70 at.%), passed this value, the increase of the sulfur content results in its organization  
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13 into rings not linked to the other glass entities.  
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18 Let us focus on the effect of the sodium addition to the glass network (with fixed S to Ge ratio)  
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20 on the basis of the Raman spectra presented in **Figure 1b**. Looking at the Raman response of  
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22 samples  $\text{Na}_0$  to  $\text{Na}_{5.5}$ , it appears that only slight changes can be observed when the sodium content  
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24 varies from 0 to 5.5 at.%. In order to highlight these spectral variations, the difference Raman  
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26 spectrum is presented in **Figure 1c**. It corresponds to the Raman spectrum of a sodium rich glass  
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28 ( $\text{Na}_{5.5}$ ) subtracted to the one of the glass without sodium ( $\text{Na}_0$ ). Multiple contributions from  $\text{GeS}_4$   
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30 tetrahedra can be observed ( $333, 375, 408$  and  $445 \text{ cm}^{-1}$ )<sup>40</sup>. The peaks at  $\sim 333 \text{ cm}^{-1}$  and  $\sim 408 \text{ cm}^{-1}$   
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32 can be ascribed to CS conformation<sup>40</sup> while the valleys at  $\sim 375 \text{ cm}^{-1}$  and  $\sim 445 \text{ cm}^{-1}$  are  
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34 representative to an Edge Sharing (ES) arrangement<sup>40</sup>. The bands corresponding to the CS  
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36 tetrahedra appear positive in the Raman difference spectra ( $\text{Na}_0$  minus  $\text{Na}_{5.5}$ ) while those related to  
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38 the ES are negative. This evolution denotes a modification of the interconnections of the germanate  
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40 network with the addition of sodium as part of the CS tetrahedral units are substituted for the ES  
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42 tetrahedra conformation. A peak at  $304 \text{ cm}^{-1}$  and a valley at around  $264 \text{ cm}^{-1}$  are present in the  
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44 difference spectra and can be attributed to the contribution of antimony glass forming units. The  
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46 addition of sodium in the vicinity of a  $\text{SbS}_3$  pyramidal units can affect the strength of the Sb-S  
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48 bonds. Therefore, the lower frequency of the band at  $264 \text{ cm}^{-1}$  for the sodium rich glasses ( $\text{Na}_{5.5}$ )  
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as compared to  $304\text{ cm}^{-1}$  without sodium ( $\text{Na}_0$ ) could be linked to weaker Sb-S bonds as part of the unit electronegativity is now distributed within the sodium coordination shell.

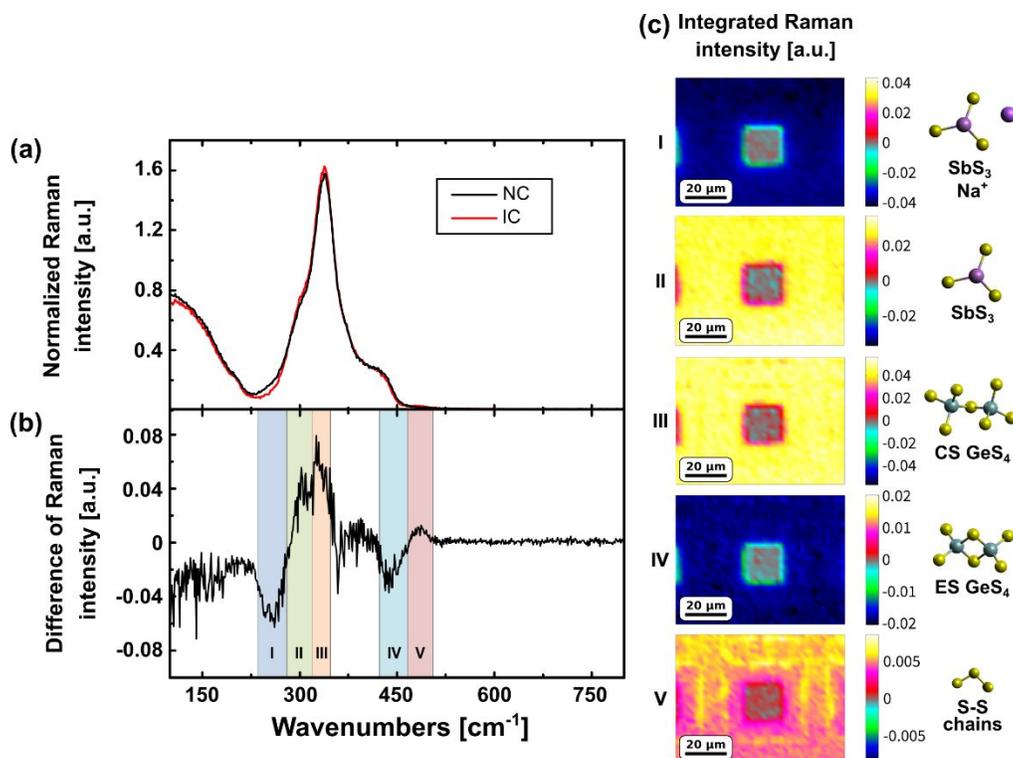


**Figure 1.** Raman spectra of the pristine ChG samples: (a) A, B, C and D glasses series with increasing S/Ge ratio, (b)  $\text{Na}_0$ ,  $\text{Na}_2$  and  $\text{Na}_{5.5}$  glasses series with increasing sodium content and constant ratio between the other components and (c) Raman difference spectra: response of  $\text{Na}_{5.5}$  subtracted to  $\text{Na}_0$  response.

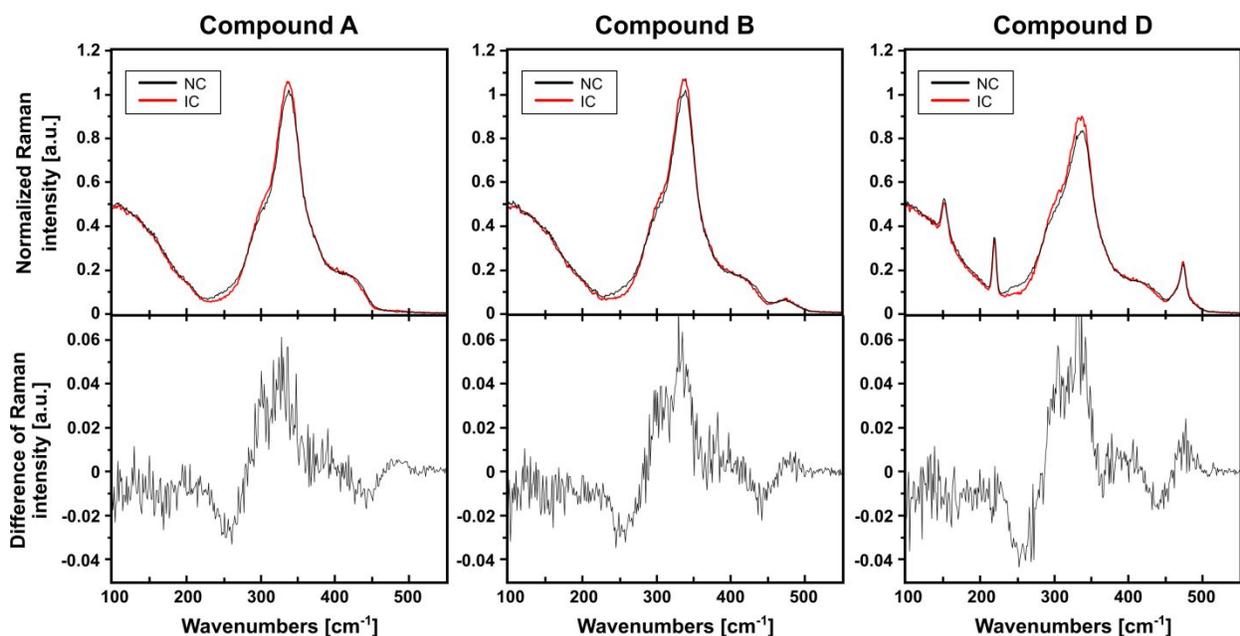
### *Spatial distribution of sodium after thermoelectrical imprinting*

The thermally poled glasses were studied by Raman spectroscopy to evidence the structural variations induced by the process. Based on previous studies,<sup>32,33</sup> a spatial control of the sodium concentration through the use of a structured electrode is expected on the post-poling surface. In **Figure 2a**, are presented the Raman spectra from the post-poling surface of the A composition. The spectra were extracted from the regions that were *in contact* (IC) with the Ti-Pt metallic areas and *not in contact* (NC) with these conductive zones (i.e. middle of the squared patterns of the electrode). The variations between the two spectra are weak, thus, in order to highlight them, the Raman difference spectrum is presented **Figure 2b**. It corresponds to the spectrum extracted from

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3 the NC region subtracted to the IC region. The resulting Raman difference spectrum closely  
4 matches the one presented in **Figure 1c** ( $\text{Na}_0$  minus  $\text{Na}_{5.5}$ ). Such similarity suggests in the IC zone  
5 the structure approaches the structure of a glass without sodium. This confirms the sodium  
6 depletion under the conductive zones of the electrode previously reported.<sup>26</sup> A similar  
7 methodology has been applied to the glasses presenting different S/Ge ratio (with same initial  
8 sodium content). The resulting spectra and Raman difference spectra are presented **Figure 3**. The  
9 ones of glass A have been reproduced for comparison with two different sulfur contents (glass  
10 samples B and D). All the Raman difference spectra show the same evolution. For all S to Ge ratio,  
11 thermal poling has the same effect: the post-poling glass in the IC zone shows a structure similar  
12 to a pristine glass without sodium. This shows the flexibility of the glass matrix which is rearranged  
13 through poling whatever the initial composition. **Figure 2c** are presented the Raman cartographies  
14 associated with selected bands of the Raman difference spectrum showing the spatial distribution  
15 of the structural rearrangements. The same pattern is observed for the distribution of all bands  
16 demonstrating the spatial coherence of the spectral variations. Furthermore, the pattern of the  
17 electrode can be recognized in these cartographies confirming that, through the use of a structured  
18 electrode, thermal poling can be successfully employed as an imprinting process. One should also  
19 notice that the imprinted patterns do not correspond exactly to the electrode design. This is due to  
20 a gradient in sodium concentration from the edge of the electrode toward the center of the non-  
21 conductive square. Such effect was accurately described in a study linked to the formation of  
22 gradient of refractive indices (GRIN) in similar glasses.<sup>26</sup>  
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**Figure 2.** (a) Raman spectra from the thermally poled A glass, the black line corresponds to the region non in contact (NC) with the conductive zone and the red line in contact (IC). (b) Raman difference spectrum where the spectrum of the NC region has been subtracted to the IC region. (c) Spatial evolution of the Raman intensity of the different bands revealed by the Raman difference spectrum and the schematic representation of the glass units they are associated with.



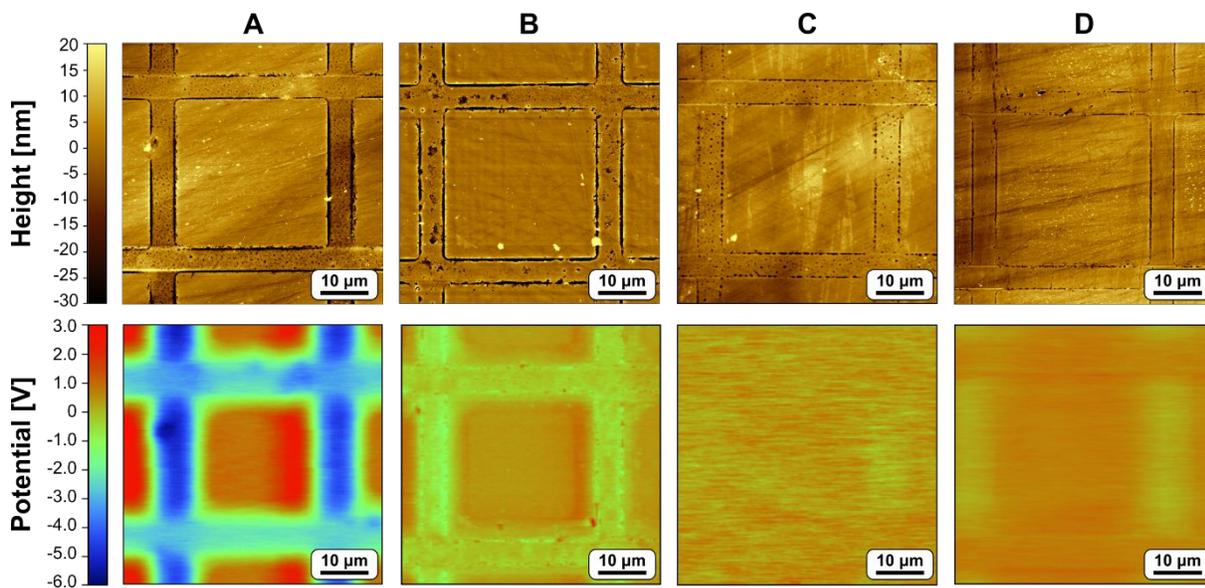
**Figure 3.** (Top) Raman spectra from the different thermally poled glasses A, B and D presenting different S/Ge ratio and same sodium content. The black lines correspond to the region not in contact (NC) with the conductive zone and the red line in contact (IC). (Bottom) Raman different spectra for each glass where the spectrum of the NC region has been subtracted from the spectrum of the IC region.

### *Patterning of the surface electrical potential*

The post-poling topography as well as the resulting surface potential magnitude and spatial distribution for all the studied glasses are presented **Figure 4** and **Figure 5**. The **Figure 4** is dedicated to the influence of the sulfur content with the series of glasses A, B, C and D (1 at. % of sodium) while in **Figure 5** the emphasis is put on the sodium content with the Na<sub>0</sub>, Na<sub>2</sub> and Na<sub>5.5</sub> glasses.

Let us focus first on the **Figure 4** with the A, B, C and D glasses. Topography variations of a maximum of 50 nm amplitude can be observed. The main changes are located at the edges of the

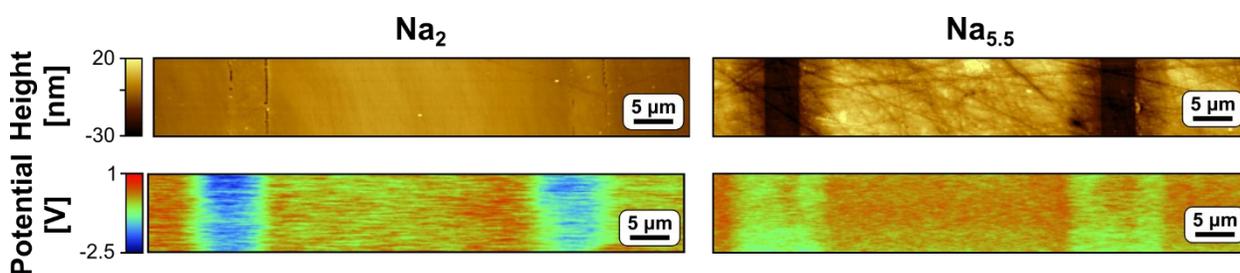
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3 electrode; the rest of the surface is rather flat for all samples with even smaller changes as the  
4 sulfur content increases. Looking at the surface potential of the poled A glass sample, one can  
5 clearly recognize the pattern of the electrode resulting in negatively charged regions when IC with  
6 the electrode and positively charged regions when NC. The region IC, as the Raman study has  
7 shown, corresponds to a sodium-depleted region. The amplitude variation of surface potential is  
8 9 V for this sample. Glass B only shows an amplitude variation of 1 V and ~0.2-0.4 V for C and  
9 D. Based on this observation the bigger the ratio S to Ge is (i.e. the more sulfur in the matrix), the  
10 lower the magnitude of the potential is: it is maximal for glass A (stoichiometric composition) and  
11 diminishes in B until the pattern can barely be observed for glasses C and D.



45 **Figure 4.** AFM measured topography (top) and KPFM measured surface potential (bottom)  
46 images of the post-poling surface of glasses A, B, C and D i.e. increasing sulfur content with fixed  
47 sodium content.  
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53 Regarding the influence of the sodium content the first comment we have to make is that when  
54 no sodium is present in the initial composition ( $\text{Na}_0$ ), the post-poling surface does not present any  
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3 topographic variation and no potential structuring is observed. For this reason, the measurements  
4 concerning this composition are not presented here. In Figure 5, can be found the post-poling  
5 topography and surface potential of glasses with higher sodium contents ( $\text{Na}_2$  and  $\text{Na}_{5.5}$ ). It appears  
6 that, for the imprinting process to be successful, the presence of alkali in the initial composition is  
7 necessary. Increasing the initial sodium content results in variation of the topography, the more  
8 the initial sodium content increases the stronger the topography changes are. The surface potential  
9 structuring can be observed for  $\text{Na}_2$  but with higher sodium contents ( $\text{Na}_{5.5}$ ), it tends to disappear.



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28 **Figure 5.** AFM measured topography (top) and KPFM measured surface potential (bottom)  
29 images of the post-poling surface of glasses  $\text{Na}_2$  and  $\text{Na}_{5.5}$  i.e. increasing sodium content with other  
30 component's ratios constant. The results obtained for  $\text{Na}_0$  (no sodium) are not presented as no  
31 effect – topology or potential – were observed in the absence of alkali cation in the initial  
32 composition.  
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## 40 41 **Discussion**

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43 The main objective of this study is to engineer a spatially varying and stable surface electrical  
44 potential on a glass optical materials that could open new opportunities in the development of  
45 multifunctional approach in inorganic glass science. Charge implantation has been successfully  
46 realized in chalcogenide materials since the 70's with the development of xerography processes  
47 notably on amorphous selenium.<sup>44,45</sup> These works have demonstrated the potential of chalcogenide  
48 materials to sustain electrical or plasma processes for the fabrication of charged surfaces.  
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3 Nevertheless, in our case, the multifunctional properties targeted require novel objectives in terms  
4 of long term stability and micrometric spatial control. As compared to a xerography approach, the  
5 thermo-electrical imprinting process is considerably different because of the nature of the charge  
6 carriers as it involves necessarily alkali cations. The Raman mapping reported has evidenced the  
7 ability of this thermo-electrical process to control the sodium spatial distribution within the  
8 chalcogenide glassy matrix at the micrometer scale for all the glass compositions studied. This  
9 structural study also points out the similarity observed between spectral variations from glasses  
10 prepared with and without sodium addition (Figure 1) and the ones measured when the sodium is  
11 depleted from the glass matrix under the electrical constrain (Figure 2). It shows the flexibility of  
12 the glass network and its ability to rearrange under a high electric field at temperatures below its  
13 glass transition. Regarding the objective of surface electrical potential patterning, the presented  
14 results have also demonstrated that by using this imprinting poling process, stable surface charges  
15 can only be obtained with sodium doped glasses which denotes the importance of cationic charge  
16 carriers to modify the surface electrical properties for these germanate chalcogenide glasses.  
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35 To analyze the surface potential patterned for this series of glasses, one should consider the  
36 respective effects of: (i) the location of the trapped charges which have to be promoted close to  
37 the surface in order to influence the surface potential and (ii) the energy level of the traps which  
38 will determine the stability of the induced surface potential.  
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45 Concerning charged surface stabilities, let's remind that research studies in the domains of  
46 polarized or charged materials have demonstrated the importance of a material disorder state to  
47 enable a slow discharge<sup>46-50</sup>. Depending on the trap energy distribution, charges can diffuse to  
48 reach a "sink" to be neutralized as described in a classical trapping model developed to describe  
49 electronic relaxation in amorphous semiconductors<sup>46,51</sup>. In addition, in reference to works done on  
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3 the photosensitivity or on thermal poling of chalcogenide glasses, it was shown that structural  
4 rearrangements induced by the optical or electrical polarization treatments are very favorable for  
5 the stability of charges as the modified structure considerably increases the energy barrier of the  
6 traps<sup>52,53</sup>.  
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11 To summarize the results of this study: (i) we have observed that the surface potential patterning  
12 is efficient only for the composition stoichiometric in sulfur, and (ii) if the sodium content  
13 increases, the electrical potential values decrease together with an enhancement of the surface  
14 topology variations. Both findings denote the importance of the glass local structure to achieve the  
15 targeted functional properties.  
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19 Concerning the effect of the sulfur content, it is not possible to conclude if the absence of surface  
20 potential observed for the rich-sulfur glasses can be linked to an implantation of trapped charges  
21 preferentially localized in the bulk or if it can be linked to a very poor stability of charges at the  
22 surface. Nevertheless, the structural study has confirmed a significant amount of S-S homopolar  
23 bonds and the presence of sulfur rings in the glassy structure, it induces a real change in the local  
24 homogeneity and polarizability of the structure which necessarily affect the distribution of trap  
25 energies and as a consequence the local transport/diffusion of charges. Under this assumption, one  
26 may expect an easier transport of the charge carriers in the poled matrix with the increase of  
27 homopolar bonds which could locally facilitate an electronic transport in the amorphous structure.  
28 In any case, it denotes the key role of the glass local molecular structure to promote stable surface  
29 charges.  
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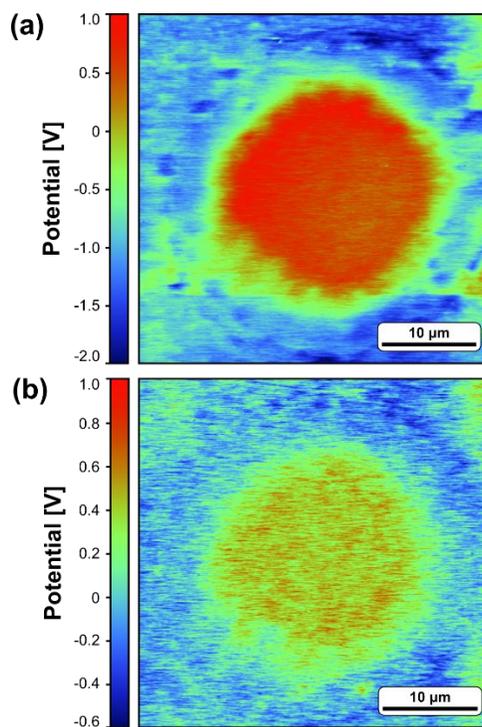
49 Focusing our analysis on the influence of the sodium content in the glass matrix, one should  
50 point out that when the amount of alkali increase the imprinting process induces non negligible  
51 surface topology variations (steps of 50 nm for 5.5 mol% of Na). This surface relief modification  
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3 is expected from previous studies on poled ionic glasses<sup>54</sup>. It denotes the presence of matter  
4 displacement, volume contraction and glass density variations induced during the alkali-depleted  
5 layer formation. In such a case, we can note a clear decrease of the patterned surface potential  
6 amplitude when the amount of charge carriers involved increase imposing large modification of  
7 the physico-chemical properties of the glassy material surface.  
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12 Furthermore, one should evoke the spatial control of charge implantation managed by this  
13 imprinting process. In regards to a future electrical/optical multifunctional approach, a spatial  
14 control of the surface potential at the micrometer scale is essential in order to suite with the sizes  
15 of photonic waveguides for example. This study has demonstrated the potential of this polarization  
16 process to achieve such objectives. To explain the surface potential patterns obtained, one should  
17 focus on the electrode configuration. : a juxtaposition of conductive and non-conductive zones (as  
18 used in this study with a metallic thin film patterned by a lithography method). Such a juxtaposition  
19 of conductive and non-conductive area allows to promote in plane field components which can  
20 promote and control surface current direction. Such effect was recently reported for the electro-  
21 optical second harmonic generation response obtained for a similar glass system<sup>32</sup>. These surface  
22 currents controlled by the electrode geometry might be at the origin of the sign control of the  
23 surface potential which allow accurate patterning at the micrometric scale, yet further studies will  
24 be required to explain this process in detail.  
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45 Finally, the question of the stability of the induced patterned surface potential with time is crucial  
46 for further integration in multifunctional systems. This temporal aspect wasn't the main objective  
47 of this work, however to justify the real interest of these polarized surfaces, is presented in figure  
48 6 the surface potential measured on a sample right after the thermoelectrical treatment and 24  
49 months later. The electrode used is constituted of a circular pattern. One can observe that, if the  
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3 amplitude variation of surface potential has significantly decreased (from 3 V to 1.6 V), the  
4 patterning of the surface potential is retained after two years. These preliminary results on the  
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6 patterning of the surface potential is retained after two years. These preliminary results on the  
7  
8 stability of the poled chalcogenide surface properties are encouraging. Some additional  
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10 characterizations are actually in progress.



36 **Figure 6.** KPFM imaging of a ChG sample after thermal poling at after being poled (a) and after  
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38 24 months of being poled (b).  
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## 41 **Conclusions**

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45 A spatial control of the surface electrical potential of chalcogenide glasses in the system Ge-Sb-S-  
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47 Na was achieved using a thermo-electrical imprinting process. The charged chalcogenide surfaces  
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49 have shown electrical potentials varying in sign within an amplitude range of 10V. The induced  
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51 patterns can be accurate at the micrometer scale and effective on large centimeter scales.  
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54 Combining data from structural characterization and Kelvin Probe Force Microscopy (KPFM), the  
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3 glass composition has been found to play a crucial role on the final electrical properties. The  
4 presence of S-S homopolar bonds and sulfur rings in glasses as well as the sodium content largely  
5 influence the induced surface potential. Such set of results open the way to create new multi-  
6 functional (optical and electrical) surfaces.  
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### 22 **Author Contributions**

23  
24 The manuscript was written through contributions of all authors. All authors have given approval  
25 to the final version of the manuscript. ‡These authors contributed equally.  
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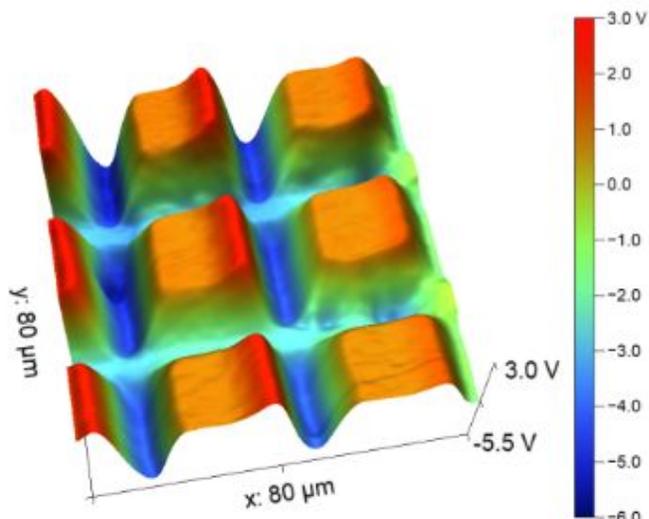
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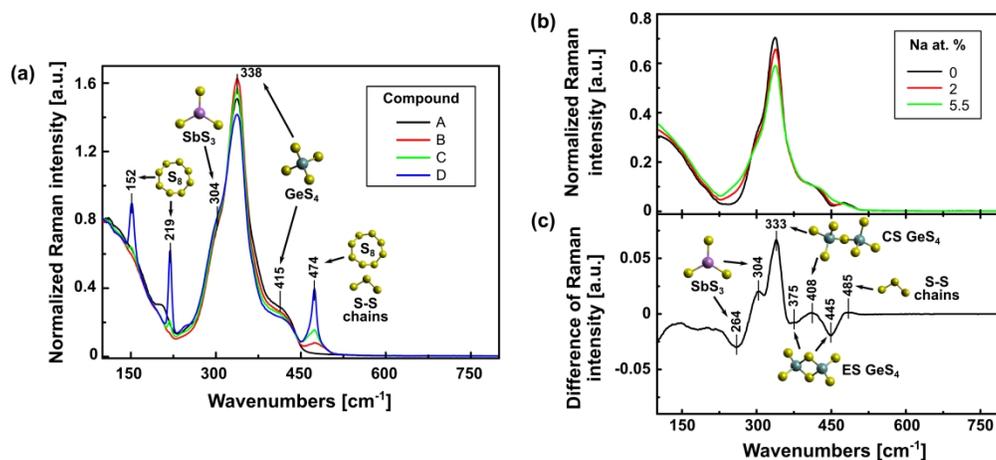


Figure 1. Raman spectra of the pristine ChG samples: (a) A, B, C and D glasses series with increasing S/Ge ratio, (b) Na0, Na2 and Na5.5 glasses series with increasing sodium content and constant ratio between the other components and (c) Raman difference spectra: response of Na5.5 subtracted to Na0 response.

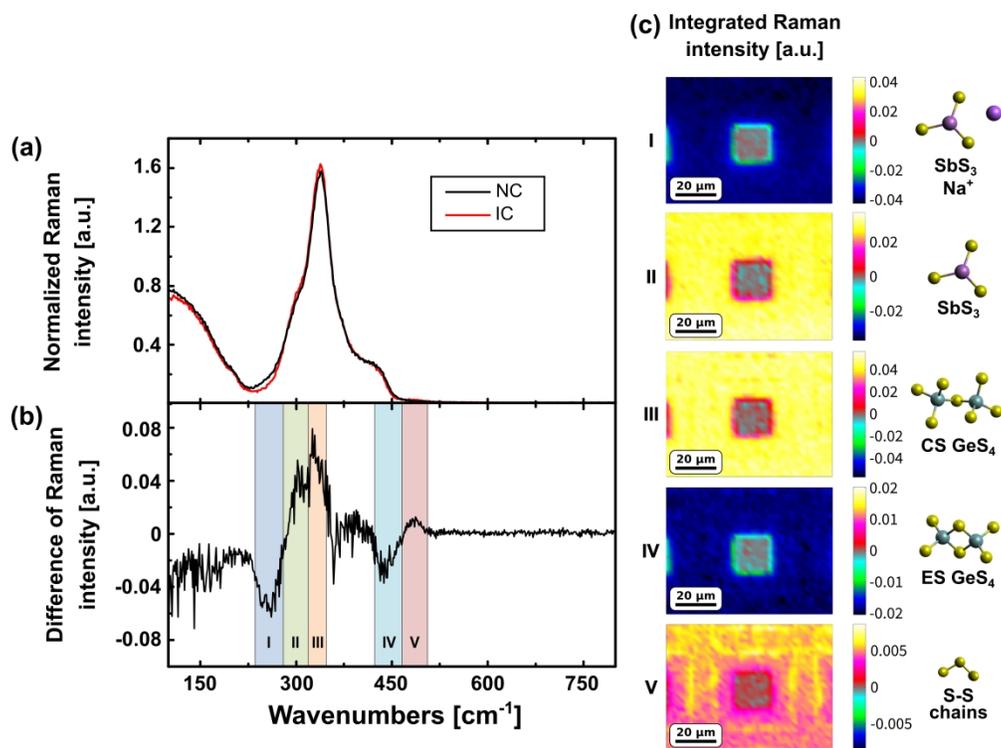


Figure 2. (a) Raman spectra from the thermally poled A glass, the black line corresponds to the region non in contact (NC) with the conductive zone and the red line in contact (IC). (b) Raman difference spectrum where the spectrum of the NC region has been subtracted to the IC region. (c) Spatial evolution of the Raman intensity of the different bands revealed by the Raman difference spectrum and the schematic representation of the glass units they are associated with.

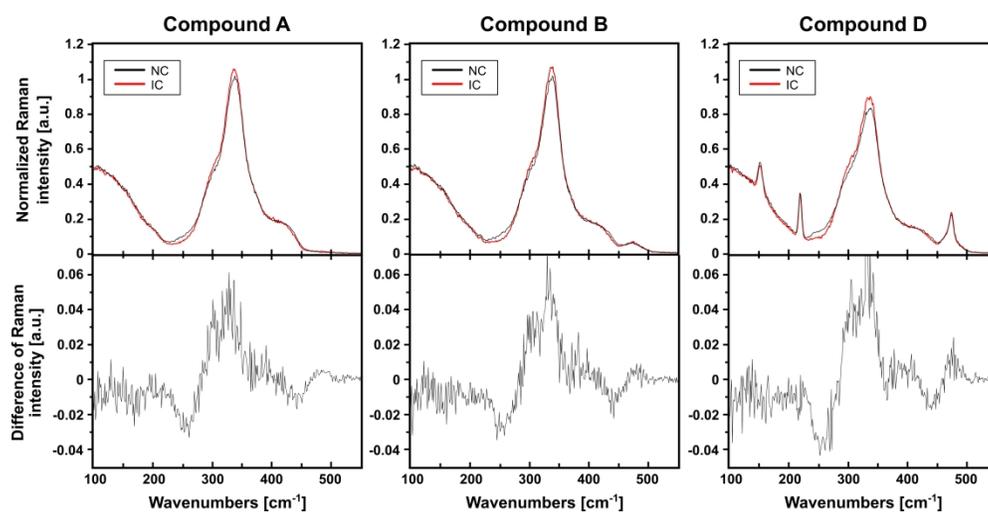


Figure 3. (Top) Raman spectra from the different thermally poled glasses A, B and D presenting different S/Ge ratio and same sodium content. The black lines correspond to the region not in contact (NC) with the conductive zone and the red line in contact (IC). (Bottom) Raman different spectra for each glass where the spectrum of the NC region has been subtracted from the spectrum of the IC region.

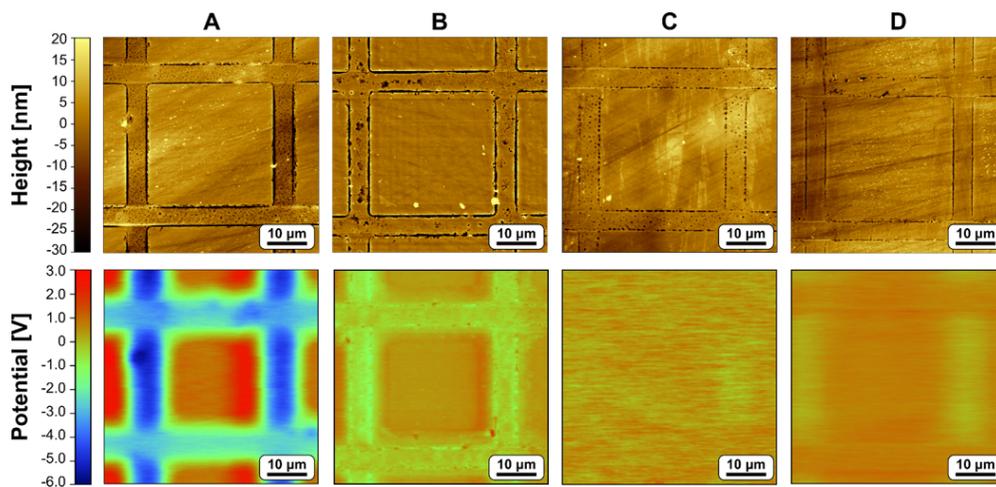


Figure 4. AFM measured topography (top) and KPFM measured surface potential (bottom) images of the post-poling surface of glasses A, B, C and D i.e. increasing sulfur content with fixed sodium content.

178x84mm (150 x 150 DPI)

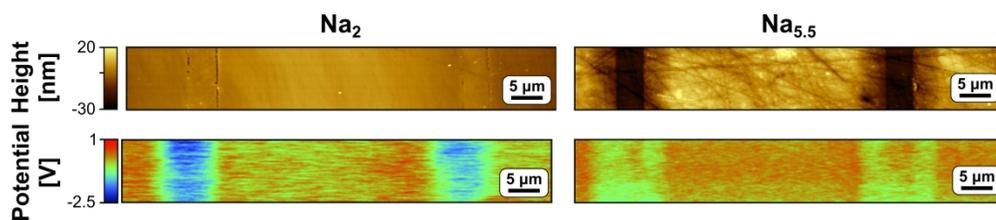


Figure 5. AFM measured topography (top) and KPFM measured surface potential (bottom) images of the post-poling surface of glasses  $\text{Na}_2$  and  $\text{Na}_{5.5}$  i.e. increasing sodium content with other component's ratios constant. The results obtained for  $\text{Na}_0$  (no sodium) are not presented as no effect – topology or potential – were observed in the absence of alkali cation in the initial composition.

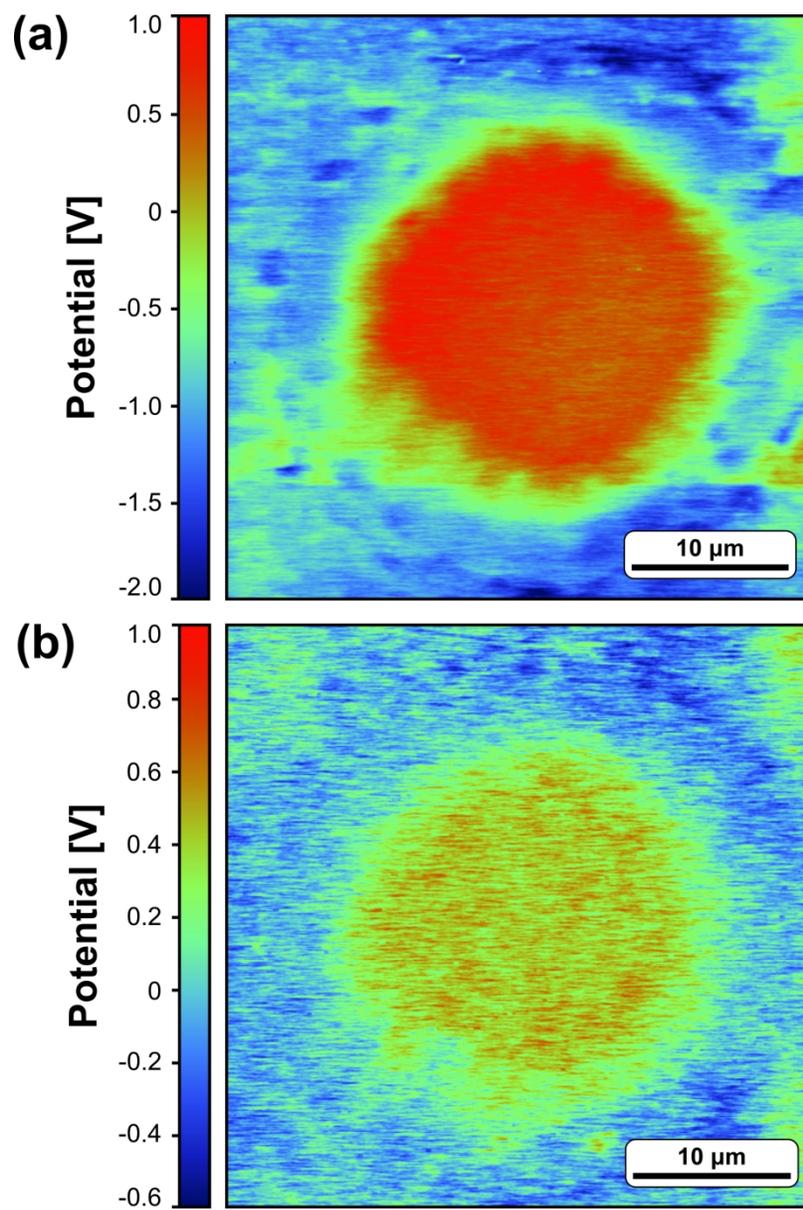


Figure 6. KPFM imaging of a thermally poled glass B using a disk pattern on the electrode measured just after the treatment (a) and after 24 months (b).