



Cite this: *Nanoscale Adv.*, 2023, 5, 3063

## Hollow @CuMgAl double layered hydrotalcites and mixed oxides with tunable textural and structural properties, and thus enhanced NH<sub>3</sub>-NO<sub>x</sub>-SCR activity

Tomasz Kondratowicz, <sup>a\*</sup>ab Ondřej Horký, <sup>a</sup> Stanislav Slang, <sup>c</sup> Lada Dubnová, <sup>a</sup> Marta Gajewska, <sup>d</sup> Lucjan Chmielarz <sup>e</sup> and Libor Čapek <sup>a\*</sup>a

Well-organized, spherical, mesoporous hollow @CuMgAl-LDHs (layered double hydroxides) are prepared by the controlled removal of the SiO<sub>2</sub> from SiO<sub>2</sub>@CuMgAl-LDH core–shell hybrids that in turn are synthesized *via* a bottom-up strategy. The materials are prepared with various Cu/Mg molar ratios (Cu/Mg = 0.05–0.50) while keeping the ratio of Cu and Mg constant, (Cu + Mg)/Al = 2. The effect of Cu doping and the silica core removal process (conducted for 4 h at 30 °C using 1 M NaOH) on the chemical composition, morphology, structure, texture and reducibility of the resulting materials are described. @CuMgAl-MOs (mixed oxides) obtained by thermal treatment of the @CuMgAl-LDHs are active and selective catalysts for the selective catalytic reduction of NO<sub>x</sub> using ammonia, and effectively operate at low temperatures. The N<sub>2</sub> yield increases with increased Cu content in the CuMgAl shell, which is associated with the easier reducibility of the Cu species incorporated into the MgAl matrix. @CuMgAl-MOs show better catalytic performance than bulk CuMgAl MOs.

Received 27th February 2023  
Accepted 26th April 2023

DOI: 10.1039/d3na00125c  
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## Introduction

NO<sub>x</sub> is primarily made up of nitric oxide (NO) and nitrogen dioxide (NO<sub>2</sub>) and belongs to a significant group of air pollutants which have a number of negative impacts on human health and the environment, including *e.g.*, photochemical smog, ozone depletion and acid rain.<sup>1</sup> The anthropogenic emission of NO<sub>x</sub> originates mainly from the stationary combustion of fossil fuels, such as that found in steel plants, power plants, cement plants and waste incineration. NO<sub>x</sub> emissions also originate from mobile sources, with a major contribution from the use of diesel engines. Considering the above, NO<sub>x</sub> emissions should be reduced by 42% for all EU member countries over the years 2020–2029.<sup>2</sup>

Up to now, the most recognized and effective technology for NO<sub>x</sub> abatement is the selective catalytic reduction of NO<sub>x</sub> using ammonia (NH<sub>3</sub>-SCR), urea, or hydrocarbons (HC-SCR).<sup>3</sup>

<sup>a</sup>University of Pardubice, Faculty of Chemical Technology, Department of Physical Chemistry, Studentská 573, 532 10 Pardubice, Czech Republic. E-mail: tomasz.kondratowicz@chem.ox.ac.uk; libor.capec@upce.cz

<sup>b</sup>Chemistry Research Laboratory, Department of Chemistry, University of Oxford, 12 Mansfield Road, Oxford, OX1 3TA, UK

<sup>c</sup>University of Pardubice, Faculty of Chemical Technology, Center of Materials and Nanotechnologies, Studentská 95, 532 10 Pardubice, Czech Republic

<sup>d</sup>Academic Centre for Materials and Nanotechnology, AGH University of Science and Technology, Mickiewicza 30, 30-059 Kraków, Poland

<sup>e</sup>Jagiellonian University, Faculty of Chemistry, Gronostajowa 2, 30 387 Kraków, Poland

Commercial NH<sub>3</sub>-SCR technology has been implemented in thermal power plants using V<sub>2</sub>O<sub>5</sub>-WO<sub>3</sub>(MoO<sub>3</sub>)/TiO<sub>2</sub> catalysts with relatively high efficiency in the temperature range of 300–400 °C.<sup>4</sup> However, exhaust gas temperatures in non-energy industries are often much lower, rendering vanadium-based SCR catalysts less efficient. Therefore, the development of catalysts with a satisfactory low-medium temperature catalytic activity window (<250 °C) and good resistance stability (*e.g.*, in the presence of SO<sub>2</sub> or HCl) is a permanent challenge for chemists and materials scientists for both economic and environmental reasons.<sup>4–6</sup> Among the many proposed systems, metal oxides, such as Cu,<sup>7,8</sup> Fe<sup>9,10</sup> and Mn,<sup>11,12</sup> are widely recognized and considered to be effective active components for low-medium temperature SCR catalysts, mainly due to their high catalytic activity, long lifetimes, wide availability and relatively low production costs. Besides the qualitative composition of the catalysts, the transition metal content also greatly influences the efficiency in the NH<sub>3</sub>-SCR process.<sup>13–15</sup> However, synthesis of multicomponent oxide catalysts by traditional preparation methods, *e.g.*, coprecipitation, hydrothermal synthesis and sol-gel methods, often leads to a heterogeneous distribution of the active components, as well as agglomeration and sintering of particles, which may significantly limit the potential of these catalysts for practical applications.<sup>1</sup>

Therefore, mixed oxides (MOs) originating from layered double hydroxides (LDHs) with different combinations of metal ions have been successfully tested as SCR catalysts since the late



1990s.<sup>1</sup> Such great interest in this class of materials is related to their unique properties, such as having excellent and uniform metal oxide dispersions, high thermal stabilities and memory effects. Moreover, their properties and surface compositions can be easily tuned during the synthesis, their preparation methods are well known, and they are environmentally friendly materials that are relatively cheap to produce. However, frequently LDH materials prepared using conventional methods have poor morphologies, such as aggregated powders composed of irregular particles of various sizes (stone-like morphologies).<sup>16</sup> This is the result of severe platelet aggregation caused by *ab*-face stacking during the drying process and it may negatively affect the porosity of the LDH materials and significantly limits their utilization.

Given the unique and important properties of pristine LDHs and MOs, various functionalization strategies have been applied to bring even more exciting performance outcomes. One such strategy is the dispersion of LDHs *via* aqueous miscible organic solvent treatment (AMOST) into thin nanosheets or *via* exfoliation into single layers.<sup>17,18</sup> Another class of LDHs modification is the generation of additional porosity, and thus a significant expansion of the specific surface area, in pristine LDHs, which enables better dispersions of the LDHs, exposes more active sites and facilitates the mass transport of reagents towards and from the surface. This goal can be achieved through alkaline etching that utilizes the amphoteric features of the  $\text{Al}^{3+}$  ion in the LDHs crystal<sup>19,20</sup> or by the introduction of a foreign substance during the synthesis of the LDH materials. The synthesis can be performed in the presence of carbon-containing precursors, such as activated carbon,<sup>21</sup> date palm ash,<sup>22</sup> and all types of ionic and non-ionic surfactant molecules,<sup>23–26</sup> which can then be removed by thermal treatment, helping to fine-tune the surface properties of the resulting mixed oxides.

In addition to the above-mentioned methods to tailor LDHs, approaches based on the immobilization of LDH on supports is a relatively new field of materials chemistry, proposed for the first time by Zhang *et al.* in 2009.<sup>27</sup> In particular, when hard templates with well-defined shapes and dimensions are employed, three-dimensional (3D) hierarchical LDH-based materials can be constructed to further improve the diffusion limitations, accessibility of the active sites, texture properties as well as the irregular shapes and sizes of the LDHs.<sup>28,29</sup> Up to now, various types of LDH-based core–shell nanostructures with a variety of cores have been reported, *e.g.*, nonporous  $\text{SiO}_2$ ,<sup>30</sup> mesoporous silicas (*e.g.*, MCM-41, MCM-48, SBA-15),<sup>31</sup>  $\text{Fe}_x\text{O}_y$ ,<sup>32</sup>  $\text{Cu}_2\text{O}$ ,<sup>33</sup>  $\text{TiO}_2$ ,<sup>34</sup> zeolites (*e.g.*, ZSM-5, Y, TS-1),<sup>35</sup> MOFs<sup>36</sup> and carbon.<sup>37</sup>

However, the current state of the literature clearly shows that  $\text{SiO}_2@\text{MgAl-LDH}$  hybrids<sup>16,30,38–44</sup> are the most explored and well-described in the literature. This is probably related to (i) the relatively fast and uncomplicated synthesis of the spherical  $\text{SiO}_2$  core with tunable control of particle size using the Stöber process,<sup>45</sup> (ii) the high stability of the  $\text{SiO}_2@\text{LDH}$  composite due to the Si–O–M covalent bonds,<sup>46</sup> as well as (iii) the MgAl phase being among the most popular variety of LDHs. A few papers have described other  $\text{SiO}_2@\text{LDH}$  core–shell systems in which the LDH consists of NiAl,<sup>39,47,48</sup> NiFe, CoFe, CoNi<sup>47</sup> and MgGa.<sup>39</sup> However, core–shell materials containing silica as a core and

more than two elements in the LDH shell have not been thoroughly examined. To date, only three structures,  $\text{SiO}_2@\text{CeMgAl-LDH}$ ,<sup>49</sup>  $\text{SiO}_2@\text{MgAlFe-LDH}$  and  $\text{SiO}_2@\text{MgAlNi-LDH}$ ,<sup>50</sup> have been described in detail.

The study focusses on hollow  $@\text{CuMgAl-LDHs}$  and hollow  $@\text{CuMgAl-MOs}$  and is a natural continuation and extension of our previous work,<sup>38</sup> in which we presented a controlled removal of the  $\text{SiO}_2$  core from a  $\text{SiO}_2@\text{MgAl-LDH}$  system to form hollow  $@\text{MgAl-LDHs}$ . Additionally, this work perfectly fills the space describing core–shell systems containing a  $\text{SiO}_2$  core and ternary LDHs as a shell.  $\text{SiO}_2@\text{CuMgAl-LDHs}$  materials with various Cu/Mg molar ratios (0.05–0.50) are synthesized while keeping the  $(\text{Cu} + \text{Mg})/\text{Al}$  molar ratio at a constant value of 2. Additionally, the conditions for the  $\text{SiO}_2$  core removal which facilitate the formation of hollow  $@\text{CuMgAl-LDH}$  are described. The effect of doping MgAl-LDH with various amounts of Cu, the implementation of the silica core removal process and the effect of thermal treatment (to form mixed oxides) on the morphology, structural, textural and redox properties of the core–shell and hollow sphere materials are described here. Moreover, the potential of such materials in environmental catalysis (selective catalytic reduction of NO with ammonia (NH<sub>3</sub>-SCR)) has also been demonstrated.

## Experimental

### Synthesis

The spherical core–shell  $\text{SiO}_2@\text{CuMgAl-LDH}$  materials (with various Cu/Mg molar ratios with a constant Cu + Mg/Al ratio of 2) were prepared according to our previous work.<sup>38</sup> The method involves the *in situ* coprecipitation of Cu, Mg and Al precursors in the presence of non-porous spherical  $\text{SiO}_2$  particles of uniform size with *ca.* 400 nm diameters. Briefly, 100 mg of spherical  $\text{SiO}_2$  particles were dispersed in 20.0 mL of deionized water using an ultrasound bath. After 1 h, 0.96 mmol of  $\text{Na}_2\text{CO}_3$  (Lach-Ner, 100.1%) was added to the milky solution and dispersed for another 10 minutes. Subsequently, to the as-prepared suspension was added dropwise (1 mL min<sup>−1</sup>) a 19.2 mL water solution containing various amounts of  $\text{Cu}(\text{NO}_3)_2 \cdot 3\text{H}_2\text{O}$  (Lach-Ner, 101.9%),  $\text{Mg}(\text{NO}_3)_2 \cdot 6\text{H}_2\text{O}$  (Lach-Ner, 100.8%), and  $\text{Al}(\text{NO}_3)_3 \cdot 9\text{H}_2\text{O}$  (Lach-Ner, 98.6%) (total metal concentration = 1.44 mmol). The pH was maintained at  $10.00 \pm 0.03$  using an aqueous NaOH solution (1 M) during titration. After 2 h of stirring (400 rpm), the obtained core–shell particles were filtered, washed with 2 L of deionized water and dried at 60 °C overnight. Dried samples are denoted as  $\text{SiO}_2@\text{CuMgAl}_x\text{-LDH}$ , where *x* expresses the nominal Cu/Mg molar ratio (0.05, 0.1, 0.2, 0.3, 0.4 and 0.5).

In the next step, the silica core was etched under static conditions by treatment with a NaOH solution (1 M) at 30 °C for 4 h. To remove remaining NaOH, the obtained particles were thoroughly washed with distilled water (2 L) and then dried at 60 °C. The obtained materials are denoted as  $@\text{CuMgAl}_x\text{-LDH}$ .

CuMgAl-based mixed oxides were obtained by thermal treatment of the LDH-like precursor in air at 450 °C for 4 h (heating rate of 5 °C min<sup>−1</sup>) in a muffle oven. Calcined materials are denoted as  $@\text{CuMgAl}_x\text{-MO}$ .



## Characterization

Quantitative energy-dispersive X-ray spectroscopy (EDX) measurements were performed on an electron microscope (LYRA3, Tescan) equipped with an EDX analyzer AZtec X-Max 20 (Oxford Instruments) at an acceleration voltage of 20 kV (an accelerating voltage possessing a signal from a depth of micrometres, which is sufficient for the studied core–shell particles with dimensions of approximately 800 nm). The samples were covered with 20 nm of carbon using a Leica EM ACE200 coater and subsequently 5 different spots with a size of  $200 \times 200 \mu\text{m}$  were measured and the obtained compositions were averaged. A scanning electron microscope (LYRA3, Tescan) was used for the surface morphology study using a 10 kV acceleration voltage. The samples were coated with a thin 18 nm layer of gold (Leica EM ACE200) prior to the scanning electron microscopy (SEM) analysis. Transmission electron microscopy (TEM) measurements were carried out on a FEI Tecnai TF20 X-TWIN (FEG) microscope equipped with an energy-dispersive X-ray spectrometer (EDAX), working at an accelerating voltage of 200 kV. Samples for the TEM observations were prepared by drop-casting on carbon-coated copper grids. The crystallographic structure of samples was determined using a Rigaku MiniFlex 600 diffractometer with a PDF-2 database and D/teX Ultra detector using Cu K $\alpha$  radiation operated at 40 kV and 15 mA. The X-ray diffraction (XRD) patterns were recorded in the range of  $2\theta = 5\text{--}80^\circ$  with a step of  $0.02^\circ$ . Nitrogen adsorption–desorption isotherms were recorded at  $-196^\circ\text{C}$  using a Micromeritics TriStar II instrument. Before the measurements, samples were degassed for 6 h under vacuum at  $130^\circ\text{C}$  or  $150^\circ\text{C}$  for the LDH and MO phases, respectively. Specific surface areas ( $S_{\text{BET}}$ ) were determined using the Brunauer–Emmett–Teller (BET) method. Total pore volumes ( $V_{\text{total}}$ ) were obtained from the amounts of nitrogen adsorbed at a relative pressure of 0.99, whereas the micropore ( $V_{\text{micro}}$ ) and mesopore ( $V_{\text{meso}}$ ) volumes were calculated using a *t*-plot and Barrett–Joyner–Halenda (BJH) models, respectively. In turn, the volume of macropores ( $V_{\text{macro}}$ ) was calculated as the difference between  $V_{\text{total}}$  and the sum of  $V_{\text{micro}}$  and  $V_{\text{meso}}$ . Temperature-programmed reduction with hydrogen ( $\text{H}_2\text{-TPR}$ ) was carried out in a flow system using a Micromeritics AutoChem II 2920 equipped with a TCD detector. Before a TPR experiment, 100 mg of a sample was outgassed in a flow of helium ( $25 \text{ mL min}^{-1}$ ) at  $450^\circ\text{C}$  and then for 30 min in an oxygen atmosphere at  $450^\circ\text{C}$ , followed by cooling to  $150^\circ\text{C}$ . At  $150^\circ\text{C}$ , the oxygen atmosphere was swapped to a helium atmosphere to flush the reactor with an inert gas and the reactor was cooled down to  $25^\circ\text{C}$ . After a freezer was added to the system, the hydrogen uptake from a stream of 5 vol% of  $\text{H}_2$  in Ar (a flow rate of  $25 \text{ mL min}^{-1}$ ) was analyzed over a temperature range of  $25\text{--}900^\circ\text{C}$  at a heating rate of  $10^\circ\text{C min}^{-1}$ . The number of redox sites was calculated from the integrals of the signal after their integration with external calibration.

## Catalytic tests

The catalysts were tested in the selective catalytic reduction of NO with ammonia ( $\text{NH}_3\text{-SCR}$ ). Catalytic studies were performed

on an experimental system consisting of a fixed-bed quartz microreactor with the outlet connected directly to the detector – a quadrupole mass spectrometer (QMS, PREVAC). The used gas mixture contained 0.25 vol% NO, 0.25 vol%  $\text{NH}_3$  and 2.5 vol%  $\text{O}_2$  diluted in pure helium (total flow rate of  $40 \text{ mL min}^{-1}$ ). Prior to the catalytic run, 100 mg of the sample (particle size in the range of  $250\text{--}315 \mu\text{m}$ ) was placed in the quartz microreactor and outgassed in the flow of pure helium at  $450^\circ\text{C}$  for 30 min. After the microreactor was cooled down to  $100^\circ\text{C}$ , the catalytic test was initiated with a linear heating rate of  $10^\circ\text{C min}^{-1}$ , over a temperature range of  $100\text{--}400^\circ\text{C}$ . Catalytic tests were performed under atmospheric pressure.

Moreover, for the most active catalyst, an isothermal long-term stability  $\text{NH}_3\text{-SCR}$  test at  $275^\circ\text{C}$  for 24 h was done with the same composition of the reaction mixture and mass of the catalyst sample. A catalytic study of the ammonia oxidation reaction was carried out using the same experimental system used in the case of the  $\text{NH}_3\text{-SCR}$  tests. Prior to the catalytic tests, the samples (100 mg, particle sizes of  $250\text{--}315 \mu\text{m}$ ) were outgassed in a flow of pure helium ( $20 \text{ mL min}^{-1}$ ) at  $450^\circ\text{C}$  for 30 min. Catalytic runs were conducted in the temperature range of  $100\text{--}450^\circ\text{C}$  with a gas mixture containing:  $[\text{NH}_3] = 0.5 \text{ vol\%}$  and  $[\text{O}_2] = 2.5 \text{ vol\%}$ , diluted in pure helium used as a balance gas (total flow rate  $40 \text{ mL min}^{-1}$ ). All catalytic experiments were performed at a constant gas hourly space velocity (GHSV) and a volumetric hourly space velocity (VHSV) of approx.  $12\,000 \text{ h}^{-1}$  and  $24\,000 \text{ cm}^3 \text{ h}^{-1} \text{ g}^{-1}$ , respectively.

## Results and discussion

### Features of the spherical $\text{SiO}_2\text{@CuMgAl-LDH}$ and $\text{@CuMgAl-LDHs}$

Table 1 gives the Cu/Mg, (Cu + Mg)/Al and the Si/(Si + Cu + Mg + Al) molar ratios of the pristine core–shell  $\text{SiO}_2\text{@CuMgAl-LDHs}$  and the hollow  $\text{@CuMgAl-LDHs}$  obtained after the leaching of the  $\text{SiO}_2$  core for 4 h. To keep a constant (Cu + Mg)/Al molar ratio, magnesium was isomorphically substituted by copper during the synthesis. However, the experimental Cu/Mg and (Cu + Mg)/Al molar ratios for the  $\text{SiO}_2\text{@CuMgAl-LDHs}$  were slightly

Table 1 Chemical composition of the  $\text{SiO}_2\text{@MgAl-LDH}$  core–shell materials before and after NaOH leaching

Sample	Cu/Mg	(Cu + Mg)/Al	Si/(Si + Cu + Mg + Al)
$\text{SiO}_2\text{@CuMgAl}_0.05\text{-LDH}$	0.07	2.04	0.52
$\text{SiO}_2\text{@CuMgAl}_0.10\text{-LDH}$	0.12	2.04	0.50
$\text{SiO}_2\text{@CuMgAl}_0.20\text{-LDH}$	0.25	2.12	0.48
$\text{SiO}_2\text{@CuMgAl}_0.30\text{-LDH}$	0.39	2.16	0.47
$\text{SiO}_2\text{@CuMgAl}_0.40\text{-LDH}$	0.50	2.18	0.47
$\text{SiO}_2\text{@CuMgAl}_0.50\text{-LDH}$	0.60	2.15	0.48
$\text{@CuMgAl}_0.05\text{-LDH}$	0.06	2.06	0.14
$\text{@CuMgAl}_0.10\text{-LDH}$	0.12	2.07	0.21
$\text{@CuMgAl}_0.20\text{-LDH}$	0.24	2.06	0.26
$\text{@CuMgAl}_0.30\text{-LDH}$	0.35	2.11	0.16
$\text{@CuMgAl}_0.40\text{-LDH}$	0.45	2.11	0.17
$\text{@CuMgAl}_0.50\text{-LDH}$	0.56	2.11	0.28

higher than the theoretical values. Similar discrepancies between the theoretical and real Cu/Mg molar ratios in LDHs have been reported previously in the literature.<sup>51–54</sup> The observed differences may result from the preferential incorporation of Cu<sup>2+</sup> ions into the crystal lattice during the growth of the LDH phase over Mg<sup>2+</sup> ions. It could be connected with the pH value, as the coprecipitation of various ions requires different pH values.<sup>55</sup> Thus, the pH value of 10.0, used in the synthesis of SiO<sub>2</sub>@CuMgAl-LDHs, could slightly favour the incorporation of Cu<sup>2+</sup> ions. It can also be explained by the preferential dissolution of the Mg<sup>2+</sup> ions over Cu<sup>2+</sup> during the washing of the solid materials with water. This hypothesis may be supported by the fact that Mg(OH)<sub>2</sub> is about 3 orders of magnitude more water-soluble than Cu(OH)<sub>2</sub> (the solubility equilibrium at 25 °C is  $5.61 \times 10^{-12}$  vs.  $1.99 \times 10^{-20}$ , respectively). Additionally, the partial dissolution of Al<sup>3+</sup> and Cu<sup>2+</sup> in NaOH solution, due to their amphoteric properties, cannot be excluded.<sup>19,20,56</sup> However, considering the relatively low NaOH concentration used during the modification, the impact may be negligible. In our previous study,<sup>38</sup> it was reported that the silica core cannot be completely removed from the SiO<sub>2</sub>@MgAl-LDHs even at long leaching times (up to 48 h) or at much stronger leaching conditions (2 M, 50 °C, 20 h). For the synthesis of @MgAl-LDHs, the optimal conditions chosen for the leaching of the SiO<sub>2</sub> core with respect to its catalytic behaviour were a 1 M NaOH solution for 4 h. Therefore, based on that, the same conditions for leaching the SiO<sub>2</sub> core of SiO<sub>2</sub>@CuMgAl-LDHs were implemented. It is a compromise between the degree of silica core removal from the core–shell system on one side, and the resulting morphology, textural, basic and catalytic properties on the other. For the pristine SiO<sub>2</sub>@CuMgAl-LDHs, the Si/(Si + Cu + Mg + Al) molar ratio was in the range of 0.47–0.52 (Table 1). After 4 h of SiO<sub>2</sub> template leaching, the Si/(Si + Cu + Mg + Al) molar ratio decreased, and it was found to range from 0.14 to 0.28. It should be emphasized that despite application of the same conditions (4 h, 1 M NaOH), it was not possible to achieve the same level of the SiO<sub>2</sub> core removal for all the @CuMgAl-LDHs (Table 1).

Fig. 1A and B show XRD patterns of the SiO<sub>2</sub>@CuMgAl-LDHs and @CuMgAl-LDHs, respectively. Also shown is the diffractogram of the reference material CuMgAl\_ref-LDH. The XRD patterns of all materials contained diffraction lines at 11.6°, 23.4°, 34.8°, 39.2°, 46.8°, 60.6° and 62.0° that could be assigned to the (0 0 3), (0 0 6), (0 1 2), (0 1 5), (0 1 8), (1 1 0) and (1 1 3) planes, respectively, typical for classical double-layered structures with the trigonal  $R\bar{3}m$  space group.<sup>57,58</sup> The shape of the diffraction lines does not vary significantly with increasing copper content (Cu/Mg molar ratio), neither in intensity nor in peak position shift. Moreover, for the pristine SiO<sub>2</sub>@CuMgAl-LDHs, a broad hump at  $2\theta \approx 22^\circ$  is distinctly visible and is related to the presence of the amorphous SiO<sub>2</sub> core in the core–shell particles.<sup>38,59</sup> After 4 h of alkali treatment, none of the positions of the LDH diffraction lines change, while their intensities slightly increased, and the broad hump almost disappears (the slight bend in the baseline between the (003) and (012) planes is still visible). This confirms a decrease in the silica content of the @CuMgAl-LDHs relative to their corresponding pristine structures. Table 2 contains the lattice parameters  $a$  ( $a = 2d_{110}$ ),  $c$  ( $c = 3d_{003}$ ), LDH basal spacing (also known as the  $d$  value or interlayer distance along the  $c$  axis, calculated from  $d(003)$ <sup>60</sup>) as well as the LDH crystallite size in the stacking direction  $D(003)$  and plane direction  $D(110)$ . The lattice parameters and LDH basal spacings are almost identical for all SiO<sub>2</sub>@CuMgAl-LDHs regardless of the LDH composition. Additionally, these parameters were maintained around the same values in the @CuMgAl-LDHs (after leaching of the SiO<sub>2</sub> core for 4 hours). The determined unit cell in the crystal lattice and the  $d$  value for the SiO<sub>2</sub>@CuMgAl-LDHs and hollow @CuMgAl-LDHs correspond to the previously reported SiO<sub>2</sub>@MgAl-LDHs and hollow @MgAl-LDHs<sup>38</sup> as well as other classic bulk double-layered structures with the similar M<sup>2+</sup>/Al<sup>3+</sup> molar ratios.<sup>60,61</sup> This may be explained by the almost identical sizes of the octahedral ionic radii of Cu<sup>2+</sup> (0.73 Å) and Mg<sup>2+</sup> (0.72 Å).<sup>62</sup> On the other hand, the calculated unit cell dimensions are slightly lower than those of various SiO<sub>2</sub>@LDH core–shell systems (LDH = MgAl, NiAl, MgGa, M<sup>2+</sup>/M<sup>3+</sup> = 3).<sup>39,40</sup> This difference may be explained by the different ionic radii in the

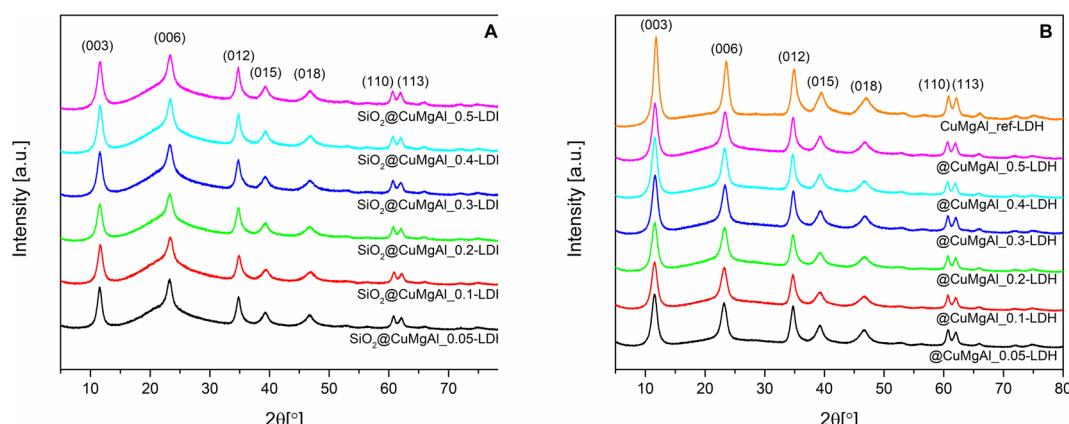


Fig. 1 XRD patterns of the SiO<sub>2</sub>@CuMgAl-LDH composites (A) before and (B) after SiO<sub>2</sub> core etching with 1 M NaOH solution. A pattern for a CuMgAl-LDH reference sample is also shown.



Table 2 Crystallographic properties of the  $\text{SiO}_2@\text{CuMgAl-LDH}$  systems before and after leaching

Sample	Lattice parameters [nm]		LDH basal spacing [nm]	Crystallite size [nm]	
	<i>a</i>	<i>c</i>		<i>D</i> (003)	<i>D</i> (110)
$\text{SiO}_2@\text{CuMgAl}_0.05\text{-LDH}$	0.305	2.291	0.764	7.8	12.7
$\text{SiO}_2@\text{CuMgAl}_0.1\text{-LDH}$	0.304	2.297	0.766	7.7	13.2
$\text{SiO}_2@\text{CuMgAl}_0.2\text{-LDH}$	0.305	2.299	0.766	7.0	13.1
$\text{SiO}_2@\text{CuMgAl}_0.3\text{-LDH}$	0.305	2.293	0.764	7.5	13.0
$\text{SiO}_2@\text{CuMgAl}_0.4\text{-LDH}$	0.305	2.295	0.765	7.1	12.3
$\text{SiO}_2@\text{CuMgAl}_0.5\text{-LDH}$	0.305	2.295	0.765	6.9	12.6
$@\text{CuMgAl}_0.05\text{-LDH}$	0.305	2.297	0.766	7.2	13.9
$@\text{CuMgAl}_0.1\text{-LDH}$	0.305	2.300	0.767	6.9	14.0
$@\text{CuMgAl}_0.2\text{-LDH}$	0.305	2.293	0.764	6.6	14.6
$@\text{CuMgAl}_0.3\text{-LDH}$	0.305	2.290	0.763	7.2	14.7
$@\text{CuMgAl}_0.4\text{-LDH}$	0.306	2.293	0.764	7.0	14.4
$@\text{CuMgAl}_0.5\text{-LDH}$	0.305	2.285	0.762	7.1	13.4

other systems as well as higher molar ratio between  $\text{M}^{2+}$  and  $\text{M}^{3+}$  in the brucite-like layers. In the case of the LDH crystallite size in the  $\text{SiO}_2@\text{CuMgAl-LDHs}$ , the calculated dimension in the stacking direction  $D$ (003) is slightly shorter than in the plane direction  $D$ (110) and are in the range of 6.9–7.8 and 12.3.1–13.2 nm, respectively. After treatment in an alkaline environment, these values remained almost unchanged in the  $@\text{CuMgAl-LDHs}$  (6.6–7.2 and 13.4–14.7 nm, respectively). Crystallite sizes that are higher in the plane direction than in the stacking direction have also been reported for other core-shell materials constructed with  $\text{SiO}_2$  cores and  $\text{NiAl}$ -,  $\text{MgGa}$ -,<sup>39</sup> and  $\text{MgAl-LDH}$ <sup>38–40</sup> shells. Previously reported  $\text{CuMgAl-LDHs}$  with a similar composition ( $\text{Cu/Mg} = 0.08$  and  $(\text{Cu} + \text{Mg})/\text{Al} = 2.03$ ) and synthesized *via* a co-precipitation method, but without the  $\text{SiO}_2$  particles, resulted in much higher crystallite sizes (75 nm).<sup>53</sup> The rapid nucleation of LDH compactly packed on the surface of the core, with simultaneous faster growth of the LDH crystal on the (110) plane than on the (003) plane, is known to ensure vertical orientation of the LDH plates onto the surface of the support.<sup>46,63</sup> Consequently, it is suggested that the  $\text{CuMgAl-LDH}$  platelets are anchored perpendicularly onto the  $\text{SiO}_2$  surface in the  $\text{SiO}_2@\text{CuMgAl-LDH}$  core–shell materials.

Regardless of the composition of the  $\text{CuMgAl-LDH}$  in the  $\text{SiO}_2@\text{CuMgAl-LDHs}$  core–shell particles, no considerable changes in their shape and size were observed using microscopic techniques. Therefore, SEM and TEM images were obtained for the material with the highest Cu content, *i.e.*,  $\text{SiO}_2@\text{CuMgAl}_0.5\text{-LDH}$  (Fig. 2A and B, respectively). The prepared core–shell structures are evenly covered by the LDH phase with the  $\text{CuMgAl-LDH}$  platelets oriented mainly perpendicularly on the  $\text{SiO}_2$  template surface. The smooth surface of the  $\text{SiO}_2$  is not visible in any case. Moreover, the material shows uniform size distribution with an average diameter of the particles of *ca.* 800 nm and a thickness of the  $\text{CuMgAl-LDH}$  shell of  $\approx 150$  nm (estimated from more than 70 particles). After partial leaching of the  $\text{SiO}_2$  core from the  $\text{SiO}_2@\text{CuMgAl-LDHs}$  for 4 h, a significant change of morphology is observed in the resulting  $@\text{CuMgAl-LDH}$  materials due to a decrease of the  $\text{Si}/(\text{Si} + \text{Cu} + \text{Mg} + \text{Al})$  molar ratio (Fig. 2C and D). The formed

hollow  $@\text{CuMgAl-LDH}$  spheres do not collapse, retaining their original spherical shape derived from the core–shell system, although a certain part of spheres is cracked with an obviously visible void inside the grains. The possible causes of these cracks were considered previously.<sup>38</sup> The most likely explanation comes from the fact that the shells could have been formed on two or more silica spheres that were close to each other (in aggregated forms), thus blocking LDH growth in the space between the balls (contact surface). In such cases, a more complex system exists resembling, *e.g.*, a raspberry-like shape, and after  $\text{SiO}_2$  leaching and/or pulverizing in a mortar, may burst, resulting in visibly broken, empty shells. Additionally, a change in the contrast between the core and the shell is visible (Fig. 2B and D) as a result of the partial removal of the  $\text{SiO}_2$  from the materials ( $\text{Si}/(\text{Si} + \text{Cu} + \text{Mg} + \text{Al})$  molar ratio decreased from 0.48 to 0.28). Moreover, the above-mentioned diversity (between the core and shell of the core–shell and hollowed particles) is slightly less visible than was the case in our last study,<sup>38</sup> where TEM images of an analogous system ( $@\text{MgAl-LDH}$  with a  $\text{Si}/(\text{Si} + \text{Mg} + \text{Al})$  molar ratio of 0.11) were studied. Additionally, a similar effect was also reported by Chen *et al.*,<sup>16</sup> who investigated the influence of the  $\text{SiO}_2@\text{MgAl-LDH}$  synthesis conditions (temperature, pH) on the formation of core–shell, yolk–shell and hollow particles. However, the authors did not provide the chemical composition of the different types of spherical particles formed but suggested that the interface between the  $\text{SiO}_2$  and LDH phases is more stable than the silica in the core.

The isotherms for the  $\text{SiO}_2@\text{CuMgAl-LDHs}$  and  $@\text{CuMgAl-LDHs}$  are shown in Fig. 3A and B, respectively, and the determined textural parameters are presented in Table 3. All  $\text{SiO}_2@\text{CuMgAl-LDH}$  core–shell particles show IVa type  $\text{N}_2$  adsorption isotherms, according to the IUPAC classification, with H3 hysteresis loops characteristic for mesoporous materials with slit-shape pores.<sup>64</sup> Regardless of the amount of copper introduced, the volume of the mesopores was similar ( $0.18\text{--}0.19 \text{ cm}^3 \text{ g}^{-1}$ ), and the total pore volume ranged from 0.32 to  $0.39 \text{ cm}^3 \text{ g}^{-1}$ . The specific surface area of the core–shell systems slightly increases with an increase in the  $\text{Cu/Mg}$  molar ratio (from 76 to  $84 \text{ m}^2 \text{ g}^{-1}$ ). This correlates with the lowest specific surface area of  $\text{SiO}_2@\text{MgAl-LDH}$



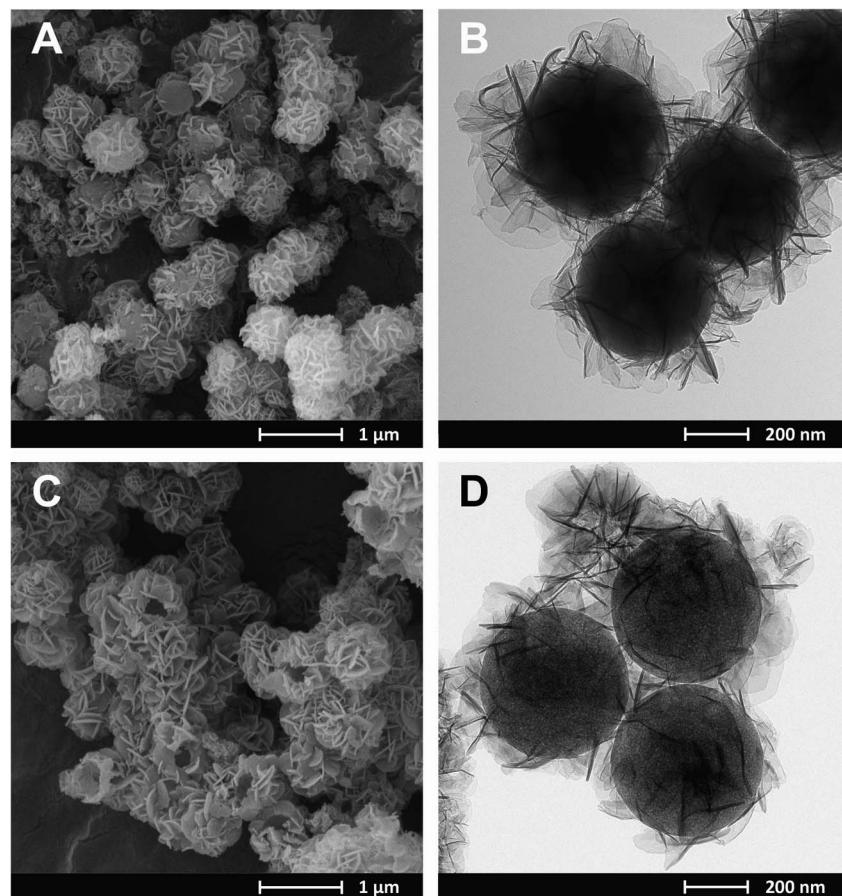


Fig. 2 SEM and TEM images of  $\text{SiO}_2$ @CuMgAl-LDH (A, B) and @CuMgAl-LDH (C, D).

(without Cu,  $65 \text{ m}^2 \text{ g}^{-1}$ ).<sup>38</sup> Considering that (i) the silica core is a non-porous solid ( $S_{\text{BET}} \approx 10 \text{ m}^2 \text{ g}^{-1}$ )<sup>38</sup> and (ii) that the influence it has on the textural properties of all the  $\text{SiO}_2$ @CuMgAl-LDH core-shell materials is similar in all cases, its impact on the above trend can be ignored. Thus, the growth of the specific surface area of the core-shell systems as the Cu/Mg molar ratio increases can only be attributed to the LDH phase. The partial elimination of silica, reached after 4 hours of etching, results in an opening of the pore structure. This effect was observed in the

measured isotherms as the significant increase in the volume of  $\text{N}_2$  adsorbed (especially noticeable at relatively low  $p/p_0$ ) and should be related to the appearance of new pores in the @CuMgAl-LDH materials. The values of  $V_{\text{total}} = 0.66\text{--}0.71 \text{ cm}^3 \text{ g}^{-1}$ ,  $V_{\text{meso}} = 0.39\text{--}0.47 \text{ cm}^3 \text{ g}^{-1}$  and  $V_{\text{macro}} = 0.21\text{--}0.28 \text{ cm}^3 \text{ g}^{-1}$  demonstrated in Table 3 confirm this hypothesis. For the sake of completeness, it should be mentioned that the micropore volume for @CuMgAl-LDHs also increased by 2 to 3 times compared to the pristine core-shell materials, but the determined volume

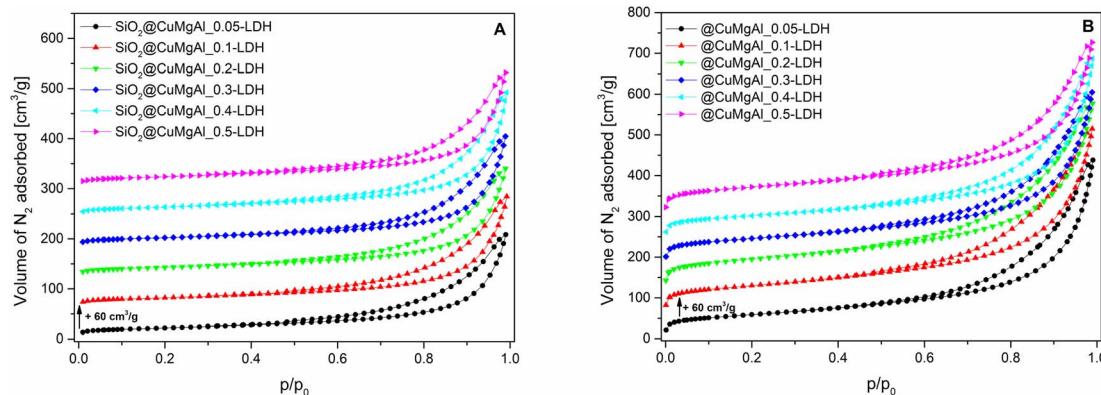


Fig. 3  $\text{N}_2$  adsorption isotherms for the  $\text{SiO}_2$ @CuMgAl-LDH composites (A) before and (B) after  $\text{SiO}_2$  core etching.

Table 3 Textural properties of the  $\text{SiO}_2$ @CuMgAl-LDH and @CuMgAl-LDH materials

Sample	$S_{\text{BET}}$ [ $\text{m}^2 \text{ g}^{-1}$ ]	$V_{\text{micro}}$ [ $\text{cm}^3 \text{ g}^{-1}$ ]	$V_{\text{meso}}$ [ $\text{cm}^3 \text{ g}^{-1}$ ]	$V_{\text{macro}}$ [ $\text{cm}^3 \text{ g}^{-1}$ ]	$V_{\text{total}}$ [ $\text{cm}^3 \text{ g}^{-1}$ ]
$\text{SiO}_2$ @CuMgAl_0.05-LDH	76	0.007	0.180	0.135	0.322
$\text{SiO}_2$ @CuMgAl_0.1-LDH	78	0.006	0.188	0.152	0.347
$\text{SiO}_2$ @CuMgAl_0.2-LDH	80	0.007	0.188	0.143	0.340
$\text{SiO}_2$ @CuMgAl_0.3-LDH	79	0.007	0.181	0.159	0.347
$\text{SiO}_2$ @CuMgAl_0.4-LDH	83	0.007	0.190	0.193	0.390
$\text{SiO}_2$ @CuMgAl_0.5-LDH	84	0.007	0.189	0.162	0.358
@CuMgAl_0.05-LDH	210	0.013	0.389	0.276	0.677
@CuMgAl_0.1-LDH	251	0.015	0.474	0.214	0.703
@CuMgAl_0.2-LDH	270	0.017	0.443	0.247	0.707
@CuMgAl_0.3-LDH	236	0.019	0.387	0.251	0.657
@CuMgAl_0.4-LDH	222	0.018	0.392	0.280	0.691
@CuMgAl_0.5-LDH	258	0.021	0.384	0.255	0.660

(0.01–0.02  $\text{cm}^3 \text{ g}^{-1}$ ) has a negligible effect on the total porosity of the studied materials. The formation of new pores affected the specific surface area, which increased to  $210\text{--}270 \text{ m}^2 \text{ g}^{-1}$ . However, an interesting correlation between  $S_{\text{BET}}$  and the chemical composition of the @CuMgAl-LDH systems was found (Fig. 4). Materials with the nominal Cu/Mg molar ratios of 0.05, 0.30 and 0.40 (group A) exhibit lower  $S_{\text{BET}}$  values, whereas the @CuMgAl-LDHs with 0.10, 0.20 and 0.50 (group B) Cu/Mg molar ratios have higher  $S_{\text{BET}}$  values. This can be explained by the effectiveness of the removal of the silica core from the  $\text{SiO}_2$ @CuMgAl-LDHs. It is clearly visible that in group A, the process of leaching the silica core was more effective than in group B (despite applying the same conditions). The lower level of silica core leaching in group B probably leads to the creation of new textured surfaces (more porous) inside the core–shell particles, which leads to an increase in  $S_{\text{BET}}$ . In group A, it can be taken that greater removal of the  $\text{SiO}_2$  core is achieved, which leads to a less porous  $\text{SiO}_2$  species.

### Spherical @CuMgAl-MO catalysts

After calcination of the LDHs at  $450^\circ\text{C}$ , all characteristic peaks of the double layered structure disappeared and XRD patterns (Fig. 5) show the presence of a poorly crystalline  $\text{MgO}$  periclase phase with three characteristic diffraction lines at  $2\theta$  of

approximately  $35.1$ ,  $43.2$  and  $62.5^\circ$  (PDF card no. 00-004-0829). For all the MOs, a slight shift of the reflections with respect to pure  $\text{MgO}$  is evident. However, irrespective of the Cu content, no highly crystalline Cu-oxide phases (e.g.,  $\text{Cu}_2\text{O}$ ,  $\text{CuO}$ ) were noted. Additionally, neither was a well-defined  $\text{CuAl}_2\text{O}_4$  spinel phase detected since the calcination temperature is too low.<sup>65–67</sup> Therefore, these observations indicate that the Cu and Al species should be well dispersed in the  $\text{MgO}$  matrix, forming a solid solution, or that well-dispersed Cu-contained phases under XRD detection limit are formed.<sup>54,62,68</sup> With an increase in the Cu/Mg molar ratio, the diffraction lines become less pronounced, indicating a decrease in the crystallinity of the MOs.<sup>69</sup> However, a large dependence between the intensity of the above-mentioned reflections and the calculated  $\text{Si}/(\text{Si} + \text{Cu} + \text{Mg} + \text{Al})$  molar ratio was not found.

The shape of the @CuMgAl-MOs  $\text{N}_2$  adsorption isotherms did not change notably in comparison to the pristine LDH-based materials (Fig. 6), implying that the mesoporous character of the hollow spheres with slit-like pores was preserved after heat treatment. In turn, all designated textural parameters (we did not consider  $V_{\text{micro}}$ , which has a negligible impact on

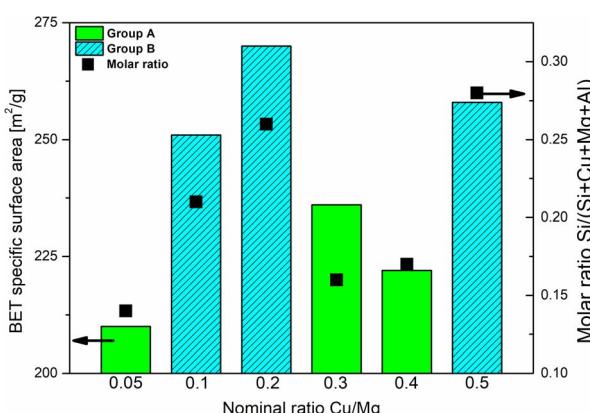


Fig. 4 Correlation between the degree of  $\text{SiO}_2$  core etching and the textural properties of the @CuMgAl-LDHs.

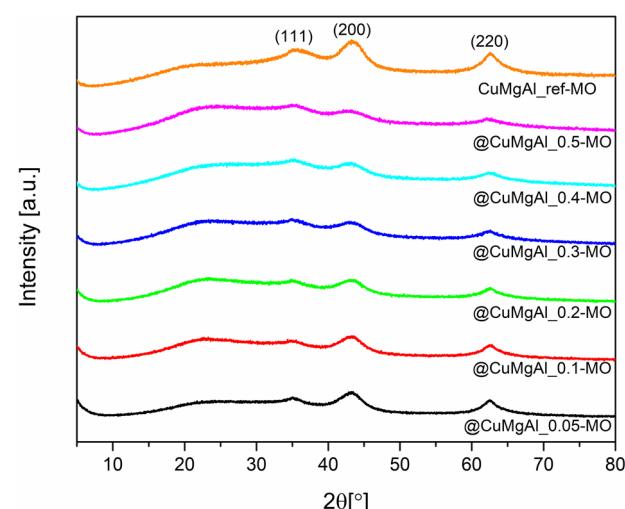
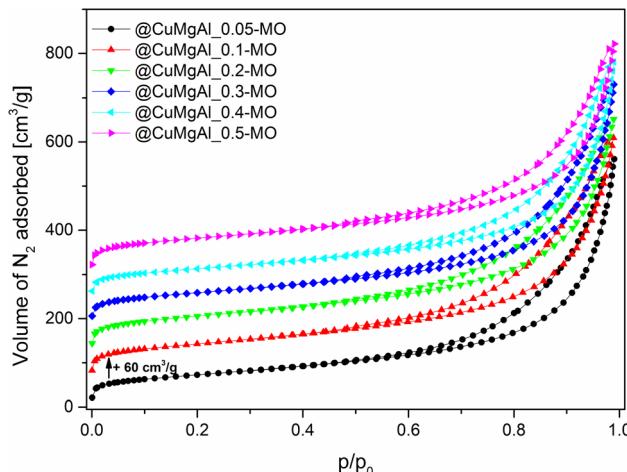


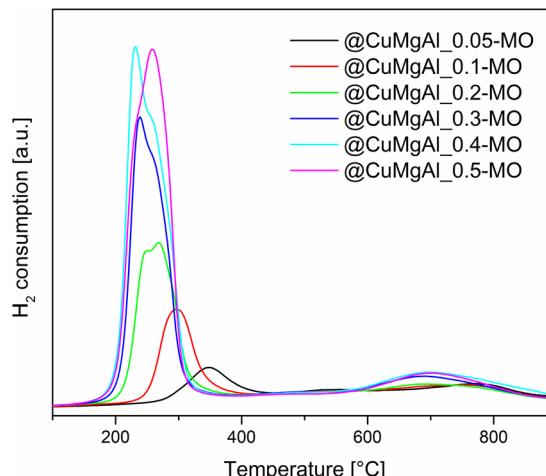
Fig. 5 XRD patterns of the @CuMgAl-MO systems and CuMgAl-MO reference catalyst.



Fig. 6  $\text{N}_2$  adsorption isotherms of the @CuMgAl-MO catalysts.

the total porosity of studied materials, Table 4) increased after transformation of the LDHs to the MOs, which is in line with the literature.<sup>70–72</sup> We noted a growth in the specific surface area by 14–24% ( $S_{\text{BET}} = 261\text{--}308 \text{ m}^2 \text{ g}^{-1}$ ), while maintaining the trend related to the presence of the silica core in the @CuMgAl materials (lower degree of  $\text{SiO}_2$  removal means higher  $S_{\text{BET}}$ ). The reported  $S_{\text{BET}}$  values are approximately two times higher than for similar CuMgAl systems with comparable compositions made at similar annealing temperatures but without spherical ordering.<sup>54,70–74</sup> Therefore, it can be concluded that the formation of mixed oxides from core–shell systems is beneficial to their resulting properties. In turn, the volume of meso- and macropores, as well as total porosity, increased after calcination by an average of 19, 31 and 23%, respectively. Considering that  $V_{\text{total}}$  was in the range of  $0.81\text{--}0.87 \text{ cm}^3 \text{ g}^{-1}$ , it can be unequivocally stated that the mesopores ( $V_{\text{meso}} = 0.45\text{--}0.54 \text{ cm}^3 \text{ g}^{-1}$ ) play a dominant role in the pore systems of the @CuMgAl-MOs, which are supplemented by macroporous voids ( $V_{\text{macro}} = 0.29\text{--}0.37 \text{ cm}^3 \text{ g}^{-1}$ ).

The reducibility of the @CuMgAl-MO samples was studied using  $\text{H}_2$ -TPR. The collected TPR profiles show two fundamental peaks of hydrogen consumption (Fig. 7). The first maximum is in the range of  $230\text{--}350 \text{ }^\circ\text{C}$  (low-thermal peak), whereas the second one is at approximately  $700 \text{ }^\circ\text{C}$  (high-thermal peak). It can be easily seen that the intensity and position of the observed low-thermal peak depends on the content of Cu in the CuMgAl mixed oxides. An increase in the Cu loading (Cu/Mg molar ratio) resulted in a higher amount of

Fig. 7  $\text{H}_2$ -TPR profiles of the @CuMgAl-MO systems pre-treated at  $450 \text{ }^\circ\text{C}$ .

$\text{H}_2$  consumed as well as a shift of the reduction process towards lower temperatures. However, different peak shapes are observed. Above a threshold of  $\text{Cu/Mg} = 0.2$ , a clear splitting or shouldering of signals is visible. Typically,  $\text{CuO}$  reduction occurs stepwise, the reduction of  $\text{Cu}^{2+}$  to  $\text{Cu}^+$  is closely followed by the reduction of  $\text{Cu}^+$  to  $\text{Cu}^0$ .<sup>75</sup> Therefore, in our case the low-thermal peak may be attributed to the complete reduction of the  $\text{Cu}^{2+}$  species incorporated in the MgAl phase to metallic Cu.<sup>76,77</sup> Additionally, the segregated  $\text{CuO}$  phase present on the surface of the MgAl oxide is rather excluded, since reduction of  $\text{CuO}$  takes place at lower temperatures.<sup>71,78</sup> The reduction temperature is connected to the strength of interaction between the Cu species and the MgAl matrix. Thus, it seems that  $\text{Cu}^{2+}$  ions present in the @CuMgAl-MO systems are strongly stabilised against reduction in a Mg-reach MgAl oxide matrix. In our study, reduction begins at temperatures as high as  $260 \text{ }^\circ\text{C}$  (for @CuMgAl\_0.05-MO) and gradually drops to  $140 \text{ }^\circ\text{C}$  (for @CuMgAl\_0.5-MO). A similar stabilisation of the Cu species deposited on pure  $\text{MgO}$  and on MgAl mixed oxides was recently reported by Basaq *et al.*<sup>77</sup> and Luggren *et al.*<sup>79</sup> Other possible explanations could be the formation of more aggregated Cu-species (still under XRD detection level) dispersed in the samples that have higher Cu content. Possibly, these more aggregated species are more easily reduced than smaller ones. In turn, the high-thermal  $\text{H}_2$ -TPR peak observed between  $600$  and  $900 \text{ }^\circ\text{C}$ , with a maximum at approximately  $700 \text{ }^\circ\text{C}$ , is associated with the reduction of  $\text{CuAl}_2\text{O}_4$  or the more complex

Table 4 Textural properties of the @CuMgAl mixed oxides

Sample	$S_{\text{BET}}$ [ $\text{m}^2 \text{ g}^{-1}$ ]	$V_{\text{micro}}$ [ $\text{cm}^3 \text{ g}^{-1}$ ]	$V_{\text{meso}}$ [ $\text{cm}^3 \text{ g}^{-1}$ ]	$V_{\text{macro}}$ [ $\text{cm}^3 \text{ g}^{-1}$ ]	$V_{\text{total}}$ [ $\text{cm}^3 \text{ g}^{-1}$ ]
@CuMgAl_0.05-MO	261	0.013	0.485	0.370	0.867
@CuMgAl_0.1-MO	294	0.015	0.536	0.297	0.849
@CuMgAl_0.2-MO	308	0.018	0.511	0.293	0.822
@CuMgAl_0.3-MO	283	0.019	0.473	0.357	0.850
@CuMgAl_0.4-MO	262	0.016	0.479	0.341	0.836
@CuMgAl_0.5-MO	296	0.021	0.449	0.335	0.806



Table 5 Redox properties of the calcined @CuMgAl materials

Sample	H <sub>2</sub> consumption [mmol g <sub>catalyst</sub> <sup>-1</sup> ]	H <sub>2</sub> consumed/H <sub>2</sub> theoretical
@CuMgAl_0.05-MO	0.28	0.42
@CuMgAl_0.1-MO	0.60	0.58
@CuMgAl_0.2-MO	1.13	0.70
@CuMgAl_0.3-MO	1.90	0.83
@CuMgAl_0.4-MO	2.59	0.96
@CuMgAl_0.5-MO	2.66	0.95

$\text{Cu}_{1-x}\text{Mg}_x\text{Al}_2\text{O}_4$  spinel phase.<sup>54,71</sup> However, it should be stressed that the studied materials were prepared by thermal treatment at 450 °C. According to Kovanda *et al.*,<sup>65</sup> a well-defined  $\text{CuAl}_2\text{O}_4$  spinel phase formed from CuMgAl LDHs only appears after calcination above 800 °C. However, it might be that the above-mentioned phase can be formed *in situ* under the conditions of the TPR experiment. On the other hand, the high-temperature peak may also reflect the reaction of H<sub>2</sub> with highly stable carbonates and/or arise from residual template species that do not decompose during thermal treatment at 450 °C. Based on the TPR curves, the total amount of redox sites was calculated from all peaks in the range of 25–900 °C (Table 5). The total number of redox sites increases with the Cu/Mg molar ratio and ranges from 0.28 to 2.66 [mmol g<sub>catalyst</sub><sup>-1</sup>]. However, the two samples with the highest nominal Cu/Mg molar ratio (0.4 and 0.5) possess almost identical H<sub>2</sub> consumption values. This can be explained by the comparable Cu content in both @CuMgAl\_0.4-MO and @CuMgAl\_0.5-MO (17.1 and 16.9 wt% of Cu, respectively), due to the fact that the latter possess more silica in the spherical system (Si/(Si + Cu + Mg + Al) is 0.22 and 0.32, respectively). The calculated values of H<sub>2</sub> consumed/H<sub>2</sub> theoretical were below 1 at all Cu/Mg molar ratios. However, an increase in this value is observed in line with the increase in Cu content (from 0.42 to 0.95). Similar findings were also reported previously for bulky CuMgAl mixed oxides with various Cu loadings<sup>54</sup> and indicate that the copper ions are only partially reduced under the applied experimental conditions. Additionally, the formation of small amounts of Cu<sub>2</sub>O during the calcination step cannot be excluded.<sup>80</sup>

### Catalytic performance of @CuMgAl-MO

The catalytic performance of the @CuMgAl-MOs as well as a CuMgAl-MO reference material was tested in the NH<sub>3</sub>-SCR reaction. The relationship between the NO conversion and N<sub>2</sub> selectivity, N<sub>2</sub> being the desired product, *vs.* temperature is presented in Fig. 8. The influence of the Cu content of the @CuMgAl-MOs on the NO conversion is very visible. The reduction of NO by ammonia in the presence of @CuMgAl-MOs started at around 150 °C and it gradually increased with increasing reaction temperature. Increasing the Cu loading in the @CuMgAl-MOs causes a significant shift of the NO conversion profiles toward lower temperatures. The maximum NO conversion level of 89% was obtained for the @CuMgAl\_0.5-MO catalyst at a temperature of 275–300 °C. The increased NO conversion of the @CuMgAl-MOs catalysts at higher Cu/Mg

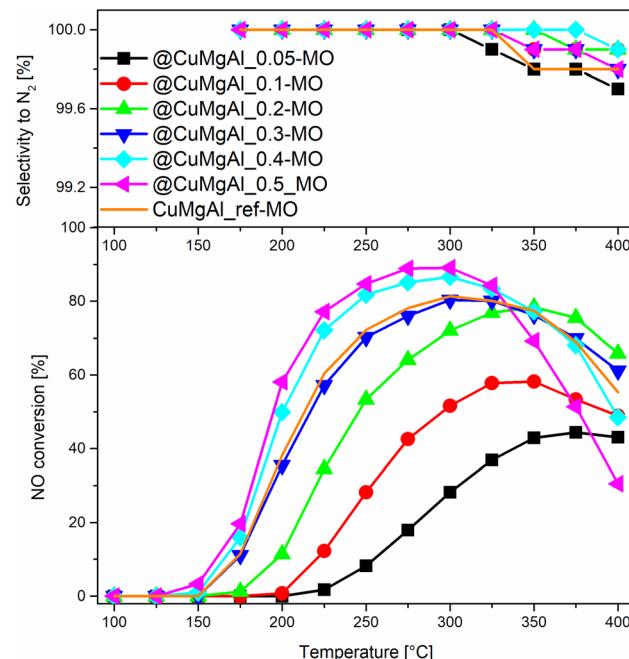


Fig. 8 Results from the NH<sub>3</sub>-SCR catalytic tests for the @CuMgAl-MOs and CuMgAl-MO reference catalyst.

molar ratios can be associated with the results of the H<sub>2</sub>-TPR measurements, where the copper species were more easily reduced at higher Cu loadings (Fig. 7). However, in the high temperature region, a decrease in the NO conversion, assigned to the side process of direct NH<sub>3</sub> oxidation by O<sub>2</sub>, was observed. Additionally, this effect is more pronounced for the MOs that demonstrate higher NO conversions. It should be noted that the N<sub>2</sub> selectivity was studied over the entire temperature range and, regardless of the composition of the @CuMgAl-MO hollow spheres, was found to be very high (over 99.7%), which indicates a high selectivity for NO to N<sub>2</sub> conversion.

To compare the catalytic behaviour of the @CuMgAl-MO hollow sphere catalysts with bulk CuMgAl-MO catalysts, a classical bulk CuMgAl-MO catalyst was prepared by coprecipitation,<sup>81</sup> and both types of catalysts were studied under the same reaction conditions. Although the CuMgAl\_ref-MO material possesses a little more Cu content than the @CuMgAl\_0.4-MO and @CuMgAl\_0.5-MO core-shell catalysts (7.0 *vs.* 5.8 and 5.7 at%, respectively), the former shows a slightly weaker catalytic performance. The NO conversion profile for CuMgAl\_ref-MO is most similar to that found for @CuMgAl\_0.3-MO. Thus, it can be concluded that the @CuMgAl-MO hollow sphere catalyst demonstrated higher NO conversion than the bulk CuMgAl-MO reference material. The lower NO conversion of bulk CuMgAl-MO may be due to the fact that the Cu-species present in this sample are harder to reduce, as concluded from the H<sub>2</sub>-TPR measurements. Moreover, the impact of its poor stone-like morphology, relatively low surface area ( $S_{\text{BET}} = 86 \text{ m}^2 \text{ g}^{-1}$ ), poor surface acidity (37.5  $\mu\text{mol g}^{-1}$ , *ca.* five times lower than for the hollow systems) and dissimilar textural properties, which may determine the accessibility of active sites for the reactants, cannot be excluded either.



The comparison of NO conversion rates achieved using the @CuMgAl-MO hollow spheres (a maximum NO conversion level of 89% was obtained using the @CuMgAl\_0.5-MO catalyst at a temperature of 275–300 °C) with the values reported in literature is dependent on the different reaction conditions. For example, Wu *et al.*<sup>82</sup> reported about 80% conversion of NO and 75% selectivity to N<sub>2</sub> at a temperature of 250 °C using a CuMgAl catalyst with a molar ratio of Cu : Mg : Al = 2 : 1 : 1. Zhang *et al.*<sup>83</sup> studied a similar system with a molar ratio of Cu : Mg : Al = 1 : 1 : 1 and reported a maximum NO conversion level of 58% at a temperature of 240 °C. In turn, Basağ *et al.*<sup>53</sup> achieved 92% conversion of NO and 96% selectivity to N<sub>2</sub> at a temperature of 325 °C, despite a much lower copper content (Cu : Mg : Al molar ratio equal to *ca.* 0.2 : 1.9 : 1.0). Chmielarz *et al.*<sup>15</sup> studied LDH-derived catalysts with Cu : Mg : Al molar ratios of 0.2 : 2.3 : 1.0, 0.35 : 2.0 : 1.0 and 0.7 : 1.8 : 1.0, which presented NO conversions at 250 °C of 85, 94 and 95%, respectively. The selectivity to nitrogen was above 90% for all these catalysts. However, a drop in the NO conversion at above 250 °C, especially for the catalysts with the highest copper loading, was observed. In the case of binary CuAl catalysts, *ca.* a 70–95% conversion of NO is achieved over temperature ranges of 200–250 °C, depending on the composition and method of catalyst preparation.<sup>84–87</sup>

The catalytic behaviour of the studied catalysts in the side process of direct ammonia oxidation by oxygen was verified by conducting additional catalytic tests. As can be seen in Fig. 9, the catalytic activity of the studied samples increased with an increase in the copper loading. However, what is very important, nearly all ammonia is oxidised to dinitrogen. Thus, residual ammonia, not converted in the NH<sub>3</sub>-SCR reaction,

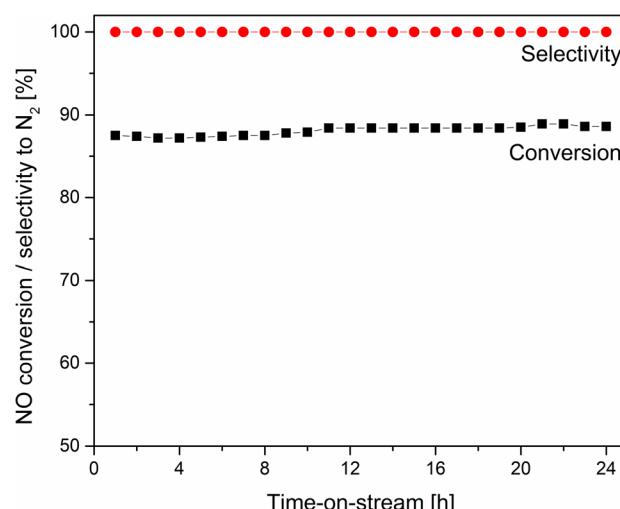


Fig. 10 Results of long-term isothermal NH<sub>3</sub>-SCR catalytic test for the @CuMgAl\_0.5-MO catalyst at 275 °C.

could be selectively oxidised to N<sub>2</sub>, which is important in the case of the possible ammonia slip.

The stability of the most active NH<sub>3</sub>-SCR catalyst (@CuMgAl\_0.5-MO) was also examined at 275 °C (Fig. 10). The collected results clearly confirm that no noticeable changes in catalytic efficiency are observed during the 24 hour catalytic run. Over the entire duration of the test, the NO conversion was stable, reaching 87.2–88.9% while the N<sub>2</sub> selectivity was 100%.

## Conclusions

The studied materials were found to be active and selective catalysts for a low-temperature selective catalytic reduction of NO<sub>x</sub> using ammonia (NH<sub>3</sub>-SCR). The activity of the catalysts increases with increasing copper content. Moreover, the temperature window for effective NO conversion is shifted to lower temperatures for the catalysts with the highest copper loadings. Thus, the copper content is very important for the formulation of effective catalysts for the low-temperature NH<sub>3</sub>-SCR process. It was shown that the temperature for copper reduction decreases with increasing copper content in the materials. The reducibility of the catalysts is, in general, correlated with their activity in the NH<sub>3</sub>-SCR process. Thus, the reducibility of the copper species seems to be very important for the activity of the studied catalysts in the low-temperature range. Comparison of the catalytic performance of the hollow @CuMgAl\_0.5\_MO sample and a reference catalyst originating from a hydrotalcite mixed metal oxide with a very similar copper loading showed significantly better activity for the hollow sample. These catalysts differ very significantly with respect to their specific surface areas (296 m<sup>2</sup> g<sup>-1</sup> for the hollow @CuMgAl\_0.5\_MO sample vs. 86 m<sup>2</sup> g<sup>-1</sup> for reference classical hydrotalcite sample). Thus, the surface accessibility of copper is possibly another very important parameter that should be considered when designing effective catalysts for the low-temperature NH<sub>3</sub>-SCR process.

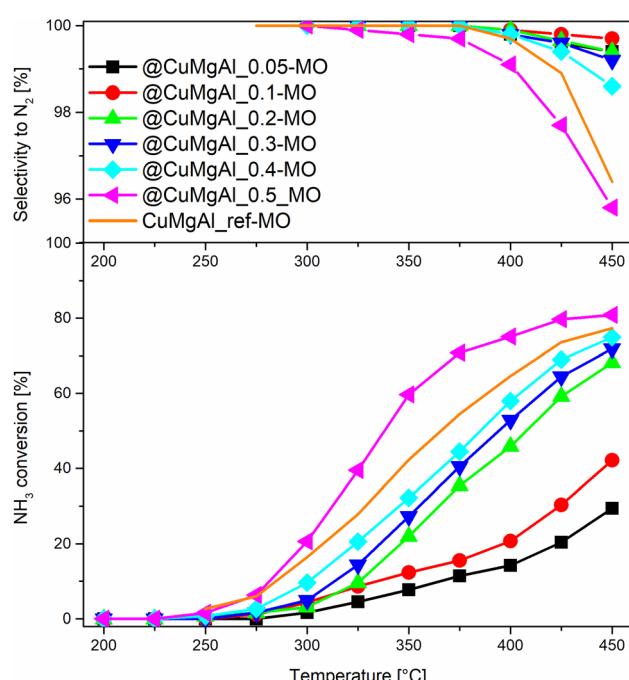


Fig. 9 Results of NH<sub>3</sub> oxidation in the presence of the @CuMgAl-MOs and CuMgAl-MO reference catalysts.



## Conflicts of interest

There are no conflicts to declare.

## Acknowledgements

The authors gratefully thank the Czech Science Foundation of the Czech Republic (Project No. 19-22978S). T. Kondratowicz also thanks European Regional Development Fund-Project “International mobility of employees of the University of Pardubice II” (No. CZ.02.2.69/0.0/0.0/18\_053/0016969”. S. Slang thanks used infrastructure (project LM2023037).

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