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## Study of the hydrogen-induced amorphization in the LaNi<sub>2.28</sub> alloy

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### Abstract:

To understand the fundamental reason leading to the poor stability in LaNi<sub>2.28</sub> hydrogen storage alloy, the cycling behavior of LaNi<sub>2.28</sub> alloy under hydrogen has been investigated. The present study describes the hydrogen-induced amorphization (HIA) of LaNi<sub>2.28</sub> alloy. The phase composition, microstructure and morphology of the phases are observed and analyzed by using X-ray diffraction and scanning electron microscopy. The crystal structure and chemical composition of the amorphized alloy are presented, and HIA is found in the sample at different hydrogenation cycle process. During the first hydrogenation cycle, the degree of HIA is indeclinable with the increased cycle. In addition, the formation processes of amorphous alloys, the occurrence of HIA and the relation between the structure of the hydrogen-induced amorphous alloy and the stability of LaNi<sub>2.28</sub> are given. Some conclusions can be first summarized:  $2\text{LaNi}_{2.28} + \text{H}_2 \rightarrow \text{LaNi}_{2.28}\text{H}_x$  (amorphous) + LaNi<sub>5</sub>H<sub>y</sub> + LaH<sub>2</sub>, and the amorphous LaNi<sub>2.28</sub>H<sub>x</sub> and LaNi<sub>5</sub>H<sub>y</sub> can desorb 50% absorption capacity, the absorption/desorption capacity is stable after the first cycle.

**Keywords:** Hydrogen storage alloys; LaNi<sub>2.28</sub>; Transmission electron microscopy;

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Hydrogen-induced amorphization; Rare earth metals.

## 1. Introduction

Because of the energy crisis and environmental issue, extensive efforts have been devoted to the development of high performance hydrogen storage alloys used as the negative electrode of the nickel-metal hydride (Ni-MH) batteries in order to significantly improve the adverse effects of consumption and dependence on fossil fuels.<sup>1-3</sup> To date, the investigated hydrogen storage alloys for the negative electrode can mainly be classified into the following categories: AB<sub>5</sub> (CaCu<sub>5</sub> type), AB<sub>3</sub> (PuNi<sub>3</sub> or CeNi<sub>3</sub> type), A<sub>2</sub>B<sub>7</sub> (Ce<sub>2</sub>Ni<sub>7</sub> type), AB<sub>2</sub> (MgCu<sub>2</sub> or MgZn<sub>2</sub> Laves phase), AB (CsCl type), Mg-based alloys and Zr-based Laves phase alloy.<sup>4-7</sup> Among them, rare earth-based AB<sub>5</sub>-type and Zr-based Laves phase hydrogen storage alloys have been widely used in the commercial Ni-MH rechargeable battery due to their long-term cycling stability and environmental friendliness.<sup>8-10</sup> However, with the rapid development of electronic devices requiring the Ni-MH rechargeable battery with higher overall performance, the limited discharge capacity of rare earth-based AB<sub>5</sub>-type (300mA·hg<sup>-1</sup>) and the difficult activation of Zr-based Laves phase alloy electrodes have become the main obstacles to extend their applications in the field of Ni-MH rechargeable battery.<sup>11,12</sup>

It has come to light that the capacity deterioration of hydrogen storage alloy used as the negative electrode is mainly associated with pulverization, oxidation or corrosion of the alloy electrode during electrochemical charge-discharge cycling.<sup>13-15</sup> Furthermore, it is worth to note that the hydrogen-induced amorphization (HIA) also plays a crucial role in destroying the hydrogen storage capacity and cycle stability of AB<sub>3</sub>-type negative electrodes.<sup>16-19</sup> Zhang *et al.*<sup>19</sup> found that the AB<sub>3</sub>-type structure consists of one-third AB<sub>5</sub> and two-third AB<sub>2</sub> subunits forming an alternating

stacking structure. For the  $\text{LaNi}_3$  crystal structure, its  $\text{LaNi}_2$  subunits absorb hydrogen resulting in forming an amorphous phase of  $\text{LaNi}_2\text{H}_x$ , while its  $\text{LaNi}_5$  subunits transfer to the crystalline phase of  $\text{LaNi}_5\text{H}_x$ , which leads to crack down of the  $\text{LaNi}_3$  crystal structure and decrease the cycle stability of the corresponding alloy. The detailed reaction process may be described as:  $3\text{LaNi}_3 \rightarrow \text{LaNi}_2\text{H}_x$  (amorphous) +  $\text{LaNi}_5\text{H}_y \rightarrow 3/5\text{LaH}_2 + 2/5\text{LaNi}_5\text{H}_z$ . As similar with the structure of  $\text{AB}_3$ -type alloy,  $\text{A}_2\text{B}_7$ -type alloy consists of both half  $\text{AB}_5$  and  $\text{AB}_2$  subunits, while  $\text{A}_5\text{B}_{19}$ -type alloy consists of three fifth  $\text{AB}_5$  and two fifth  $\text{AB}_2$  subunits. It was determined that the structure stability of  $\text{AB}_2$  subunits determines on the hydrogen storage properties of  $\text{AB}_3$ ,  $\text{A}_2\text{B}_7$  and  $\text{A}_5\text{B}_{19}$ -type alloy. With increasing the atomic proportion of B presented element among  $\text{AB}_3$ ,  $\text{A}_2\text{B}_7$  and  $\text{A}_5\text{B}_{19}$ -type alloy, the content of  $\text{AB}_2$  subunits would decline resulting in improving in the cycle stability but decreasing in the theoretical hydrogen storage and electrochemical capacity. To date, the HIA has been discovered in a great numbers of intermetallics including  $\text{A}_3\text{B}$ ,  $\text{A}_2\text{B}$ ,  $\text{AB}$ ,  $\text{AB}_2$ ,  $\text{AB}_3$ ,  $\text{A}_2\text{B}_7$ -type alloy and so on.<sup>16-23</sup>  $\text{LaNi}_{2.28}$  ( $\text{La}_7\text{Ni}_{16}$ ) alloy contains less Ni as compared with  $\text{LaNi}_3$  alloy, which means  $\text{LaNi}_{2.28}$  alloy could be a promising candidate as the negative electrode for Ni-MH rechargeable batteries due to its high theoretical hydrogen storage capacity. Therefore, it is necessary to investigate whether  $\text{LaNi}_{2.28}$  alloy exists HIA phenomenon during the absorption-desorption cycle process and the effects of HIA on the hydrogen storage capacity of  $\text{LaNi}_{2.28}$  alloy if HIA exists. Unfortunately, so far the systematic investigation about HIA in  $\text{LaNi}_{2.28}$  alloy have not been conducted.<sup>24-27</sup>

In order to a better understanding the underlying mechanism of why and how to avoid or lessen the HIA in  $\text{LaNi}_{2.28}$  alloy for improving the cycling stability of hydrogen storage alloys as the negative electrode material, in this work, the HIA process in  $\text{LaNi}_{2.28}$  alloy evolving during cycling

under hydrogen and the relationship between HIA and the microstructure of  $\text{LaNi}_{2.28}$  alloy were investigated by X-ray diffraction (XRD), scanning electron microscope (SEM) coupled with energy-dispersive spectroscopy (EDS), pressure-composition-temperature (PCT) and transmission electron microscopy (TEM) measurements.

## 2. Experimental

Alloy ingots with nominal composition  $\text{LaNi}_{2.28}$  were prepared by arc-melting the mixtures of pure La (99.9 mass%) and Ni (99.9 mass%) in a high purity argon atmosphere. To assure the homogeneity of the prepared alloy, the alloy ingots were turned over and re-melted at least three times. According to the phase diagram of La-Ni binary alloy, the annealing treatment was set at 700 °C in a bid to obtain the maximum quantity of  $\text{LaNi}_{2.28}$  phases. Consequently, the alloy samples sealed inside the quartz tubes were annealed at 700 °C for 24 h in a high-purity argon atmosphere. After completing the holding time at 700 °C, the annealed quartz tubes were immediately quenched in ice water to achieve relatively larger cooling speed resulting in more the tetragonal  $\text{LaNi}_{2.28}$  phases. To achieve fine powders from the annealed alloy samples, the annealed alloys were taken to conduct the mechanical ball milling by using a high-energy Spex mill (QM-3B) at the rate of 1200 rpm for 30 min. About 1.5 g of the annealed alloy sample was load into the stainless milling vial with a ball to powder weight ratio of 20:1 in a high-purity argon-filled glovebox ( $\text{H}_2\text{O}$ : <10 ppm;  $\text{O}_2$ : <10 ppm). Consequently, the alloy samples were mechanically ground into the powders with an average 35  $\mu\text{m}$  diameter. Herein, it worth to note that the steel vial should rest for 5 min after every milling 10 min during the milling process in order to prevent the temperature increasing inside the steel vial.

X-ray diffraction (XRD) measurements were performed to characterize the amorphous nature

and phase composition of as annealed  $\text{LaNi}_{2.28}$  alloys by using a Rigaku D/Max diffractometer with  $\text{Cu K}_\alpha$  radiation ( $\lambda=1.5406\times 10^{-10}$  m). The morphologies of as annealed samples were observed by using a Cambridge S360 model scanning electron microscope (SEM) coupled with energy-dispersive spectroscopy (EDS). The hydriding/dehydriding properties of the  $\text{LaNi}_{2.28}$  alloy samples were measured by using a pressure-composition-temperature (PCT) apparatus (Beijing Nonferrous Metal Research Institute, China). The details of the apparatus were given in previous reports<sup>28,29</sup>. Typically, 0.5 g powder sample was loaded into the vessel, and then the measuring system was pumped to a hard vacuum, then the sample was heated up to 120 °C at a 6 °C/min heating rate under 0.1 atm. Following the completion of the first absorption process, the sample was subjected to rehydrogenation at 120 °C under 3.5 MPa hydrogen pressure. Subsequently, the rehydrogenated sample was dehydrogenated at 120 °C under 0.1 MPa pressure to finish one cycle. The variation of phase composition with cycling process was examined by XRD and transmission electron microscopy (TEM).

### 3. Results and Discussion

#### 3.1 Structural characteristics

Fig. 1 shows the XRD pattern of  $\text{LaNi}_{2.28}$  alloy annealed at 700 °C for 24 h. As can be seen in Fig. 1, the annealed sample mainly consists of  $\text{LaNi}_{2.28}$  phase but still contain little  $\text{LaNi}_3$  second phase. In order to further prove the phase composition of the annealed  $\text{LaNi}_{2.28}$  alloy, Fig. 2a exhibits the back-scattered SEM image of the annealed  $\text{LaNi}_{2.28}$  alloys annealed at 700 °C for 24 h. It is obvious that the overall image is divided into two regions corresponding to grey and black region. Through EDS measurement for these two regions, it can be concluded that the grey region corresponds to  $\text{LaNi}_{2.28}$  phase and the black region should be  $\text{LaNi}_3$ , as shown in Fig. 2b and Fig. 2c.

Through analyzing the phase diagram of La-Ni alloy,<sup>30</sup> the LaNi<sub>2</sub> phase could be obtained by the peritectic reaction under equilibrium solidification, which means that it is difficult to obtain the single LaNi<sub>2</sub> phase. Moreover, Klimyenko *et al.*<sup>31</sup> found that La and Ni atoms are easier to be formed into LaNi<sub>2.28</sub> phase when the atomic ratio between La and Ni is 1:2, which is mainly because the atomic radius ratio between La and Ni (1.506) exists a relative greater difference with the ideal the atomic radius ratio of the Laves phase (1.225). Therefore, it is easier to be formed LaNi<sub>2.28</sub> phase when the atomic ratio between La and Ni is 1:2, which can explain why more LaNi<sub>2.28</sub> phases appear in Fig. 1.

### 3.2 Hydrogen absorption-desorption properties

Previous report on LaNi<sub>2</sub> found that the LaNi<sub>2</sub> alloy could achieve the equilibrrious compound corresponding to LaNi<sub>2</sub>-H<sub>2</sub> when the LaNi<sub>2</sub> alloy occurs the absorption-desorption cycle.<sup>32</sup> However, so far there is no detailed relation between absorption-desorption cycle and LaNi<sub>2.28</sub> alloy. Herein, the hydrogenation properties of LaNi<sub>2.28</sub> alloy were investigated to understand the effects of HIA on the hydrogenation properties of LaNi<sub>2.28</sub> alloy. Fig. 3a shows the hydrogen absorption curves of the as annealed LaNi<sub>2.28</sub> alloy at 120 °C under hydrogen pressure of 3.5 MPa without any activation process. As shown in Fig. 3, the as annealed LaNi<sub>2.28</sub> alloy can absorb 1.6 wt.% hydrogen within 0.5 h in the above mentioned absorption condition. As compared with the traditional LaNi<sub>3</sub> alloy,<sup>16-18</sup> the LaNi<sub>2.28</sub> alloy exhibits higher hydrogen absorption behavior at relatively low temperatures ( $\leq 150$  °C). However, the cyclic stability of LaNi<sub>2.28</sub> alloy is very poor as compared with other hydrogen storage alloys. It is necessary to explore the reason for the poor cyclic stability. The desorption kinetics of LaNi<sub>2.28</sub> was measured at 120 °C at an initial hydrogen pressure of 3.5 MPa without any activation process. As can be seen in Fig. 3b, the LaNi<sub>2.28</sub> shows a very sluggish

hydrogen desorption kinetics. The  $\text{LaNi}_{2.28}$  only desorbed about 1.0 wt. % of hydrogen at 120 °C within 1 h, indicating that the desorption kinetics of  $\text{LaNi}_{2.28}$  is worse than its absorption kinetics.

Fig.4 shows the isothermal absorption-desorption curves of  $\text{LaNi}_{2.28}$  alloy from the first to fourth cycle at 120 °C. Before measuring the second cycle, the sample was dehydrogenated at 120 °C for 0.5 h. It should be noted that 1.6 wt. % hydrogen remains inside the alloy sample after the first absorption process. The sloping plateau appears and the hydrogen absorption obviously decreases in the second absorption-desorption process, suggesting that the isothermal absorption-desorption curves of the second cycle is remarkably different from the first one. There is a distinct decrease of hydrogen absorption existing in absorption-desorption cycle of  $\text{LaNi}_{2.28}$  alloy, indicating the formation of amorphous hydrides during the cycling process.

### 3.3 Structural characteristics after hydriding/dehydriding

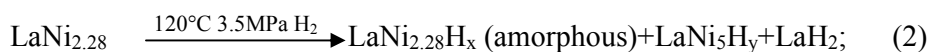
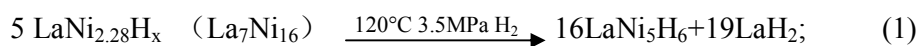
In order to understand the change of the structure and morphology of  $\text{LaNi}_{2.28}$  alloy after absorption-desorption cycles, the as-annealed alloys after hydriding/dehydriding were observed by SEM and the results are illustrated in Fig. 5. Comparing SEM images of Fig. 5 (a), (b), (c) and (d), it can be found that the change of granule sizes of the alloy samples is not obvious. This result shows that the pulverization basically occur in the process of hydriding/dehydriding, indicating that the HIA of the  $\text{LaNi}_{2.28}$  alloy are main reasons leading to the efficacy loss of the alloy hydrogenation. The formation of amorphous hydrides inside of the  $\text{LaNi}_{2.28}$  alloy impedes the diffusion of hydrogen atoms and decreases the dynamic property in the process of hydrogen absorption and desorption.<sup>33</sup> Therefore, it can be concluded that the most important approach of enhancing the cycle life of the La-Ni system ( $\text{LaNi}_{2.28}$ -type) hydrogen storage electrode alloys is to improve their anti-amorphization.



Fig. 6(a) shows the XRD pattern of as-annealed alloy after the first absorption. As compared with Fig. 1,  $\text{LaNi}_{2.28}$  and  $\text{LaNi}_3$  phases disappear and the main phases are  $\text{LaNi}_5\text{H}_y$  and  $\text{LaH}_2$ . Meanwhile, a small amount of  $\text{LaNi}_5$  can also be observed, and the diffraction peaks are comparatively broadened. Combined with the analysis of TEM (Fig. 7), it can be observed that a few amorphous phases are formed. Moreover, the diffraction rings in SAED can be indexed with crystal planes of (111) for  $\text{LaNi}_5\text{H}_y$ , (111) for  $\text{LaNi}_5$  and (101) for  $\text{LaH}_2$ , which is in good agreement with XRD analysis. Fig.6 (b), (c) and (d) show the XRD patterns of alloy after first, fourth and tenth cycles, respectively. There is no diffraction peaks observed, indicating that the phases exist in the form of amorphous state. Combined with the TEM analysis (Fig. 7 (b, c, and d)), it can be concluded that the major phases are amorphous state, and minor crystal phases corresponding to  $\text{LaNi}_5$  and  $\text{LaH}_2$  can be seen, which is verified by SAED analysis.

It can be seen from Fig. 6 (a) and Fig. 7(a) that only partial amorphous phases can be found after absorption. The disproportionation reaction happened between  $\text{LaNi}_{2.28}$  and  $\text{LaNi}_3$  following  $\text{LaH}_2$  and  $\text{LaNi}_5$  formed. The major phases exist in the form of amorphous state after the first absorption, while the minor phases are  $\text{LaH}_2$  and  $\text{LaNi}_5$  crystalline phases, which is contradictory that  $\text{LaH}_2$  and  $\text{LaNi}_5$  are in amorphous state after desorption. So far this phenomenon has not been found in other literatures. Meanwhile, this phenomenon is not accord with the process of HIA, and it can be related with the process of preparing samples. The first absorption was done at 120 °C under 3.5 MPa hydrogen pressure, the XRD and TEM measurements were done after the sample cooled in electric furnace from 120 °C to room temperature, that is to say, the sample was in amorphous state after the first absorption, but the crystallization process for samples results from cooling in the electric furnace. For the samples after the first cycle, the XRD measurement was

done after the sample cooled in air from 120 °C to room temperature. It can be seen from Fig. 4 and Fig. 5, the absorption capacity is about 1.6 wt% for the first absorption, but the desorption capacity is only 0.8 wt% for the first desorption, which indicates that disproportionation reaction happens during the absorption process following  $\text{LaH}_2$  and  $\text{LaNi}_5\text{H}_6$  formed according to the reaction formula (1). As a support for above, the precipitation of  $\text{LaH}_2$  and  $\text{LaNi}_5\text{H}_z$  from  $\text{LaNi}_2\text{H}_x$  is indeed observed by Chung and Lee.<sup>34</sup> According to Fig. 7, after first, fourth and tenth cycles at 120 °C, the SADP of  $\text{LaNi}_5$  is pronounced.  $\text{LaH}_2$  is very stable and can not desorb hydrogen at 120 °C, resulting in the decrease of desorption capacity. According to the reaction formula (1), if all  $\text{La}_7\text{Ni}_{16}$  ( $\text{LaNi}_{2.28}$ ) participates in the disproportionation reaction, the desorption capacity of the formed  $\text{LaNi}_5\text{H}_6$  is 71.6% of the total absorption capacity, but the actual desorption capacity is 50% of the total absorption capacity, which indicates that the disproportionation reaction do not happen completely. After the absorption, amorphous  $\text{LaNi}_{2.28}$ ,  $\text{LaNi}_5\text{H}_6$  and  $\text{LaH}_2$  are formed as shown in the reaction formula (2). The absorption and desorption capacity are 0.8 wt% from the second to fourth cycle, indicating that the amount of  $\text{LaH}_2$  is stable after the first cycle and the disproportionation reaction is complete during the first cycle.



#### 4. Conclusions

In the present work, the hydrogen storage behavior of  $\text{LaNi}_{2.28}$  intermetallic is systematically studied after absorption/desorption cycling under hydrogen gas, where they undergo a degree of amorphization. The hydrogen-induced amorphization (HIA) is mainly responsible for the poor reversibility of the  $\text{LaNi}_{2.28}$  intermetallic. The following conclusions can be drawn:

The hydrogen-induced amorphization (HIA) is mainly responsible for the poor reversibility of the  $\text{LaNi}_{2.28}$  alloy. The HIA mainly occurs after the first cycle. With the increase of the number of cycles, the degree of HIA is not intensified. For the  $\text{LaNi}_{2.28}$  alloy, the hydrogen absorption-desorption of its  $\text{LaNi}_{2.28}$  phase induces the local lattice to become a disorderly one, leading to the amorphous phase formation. At the same time, the  $\text{LaNi}_5\text{H}_x$  crystallites formed by the  $\text{LaNi}_5$  subunit start to precipitate. Combining these with the XRD results in Fig. 6, it describes what happened in the  $\text{LaNi}_{2.28}$  alloy: after the first absorption, amorphous  $\text{LaNi}_{2.28}\text{H}_x$  formed, and it formed  $\text{LaNi}_5\text{H}_y$  and  $\text{LaH}_2$  by the disproportionation reaction, amorphous  $\text{LaNi}_{2.28}\text{H}_x$  and  $\text{LaNi}_5\text{H}_y$  can desorb 50% absorption capacity, the absorption/desorption capacity is stable after the first cycle.

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**Figure captions**

**Fig.1** XRD pattern of LaNi<sub>2.28</sub> alloy annealed at 700 °C for 24 h..

**Fig.2** (a) back-scattered SEM image of the annealed LaNi<sub>2.28</sub> alloys annealed at 700 °C for 24 h; Energy dispersive spectroscopy (EDS) results of (b) grey region A and (c) black region B.

**Fig.3** The first hydrogen absorption-desorption curves of LaNi<sub>2.28</sub> at 120 °C under a hydrogen pressure of 3.5Mpa: (a) hydrogen absorption and (b) hydrogen desorption.

**Fig.4** The hydrogen capacity of pre-annealed LaNi<sub>2.28</sub> alloy during 4 hydriding/dehydriding cycles at 120 °C.

**Fig.5** SEM images of as-cast and hydriding/dehydriding of the microstructure: (a) as-cast; (b) 1 time hydriding/dehydriding; (c) 4 hydrogen hydriding/dehydriding cycles and (d) 10 hydriding/dehydriding cycles.

**Fig.6** XRD patterns of the pre-annealed LaNi<sub>2.28</sub> alloy cycled under 3.5 MPa H<sub>2</sub> after (a) 1 time hydrogen absorption; (b) 1 time hydrogen desorption; (c) 4 times hydrogen desorption and (d) 10 times hydrogen desorption.

**Fig.7** TEM images and selected area diffraction patterns of the pre-annealed LaNi<sub>2.28</sub> alloy after (a) 1 time hydrogen absorption; (b) 1 time hydrogen desorption; (c) 4 times hydrogen desorption and (d) 10 times hydrogen desorption.

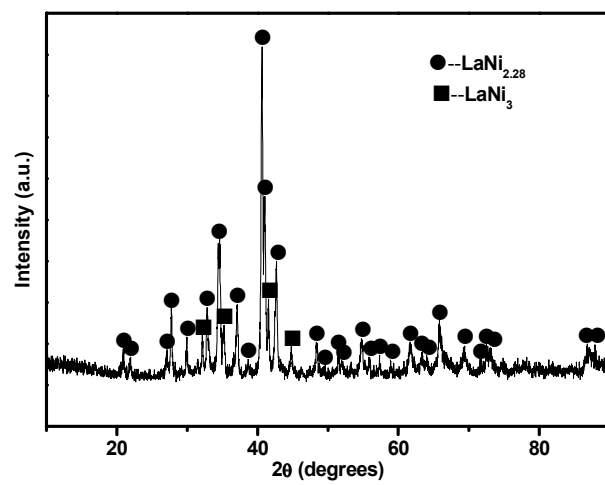


Fig.1

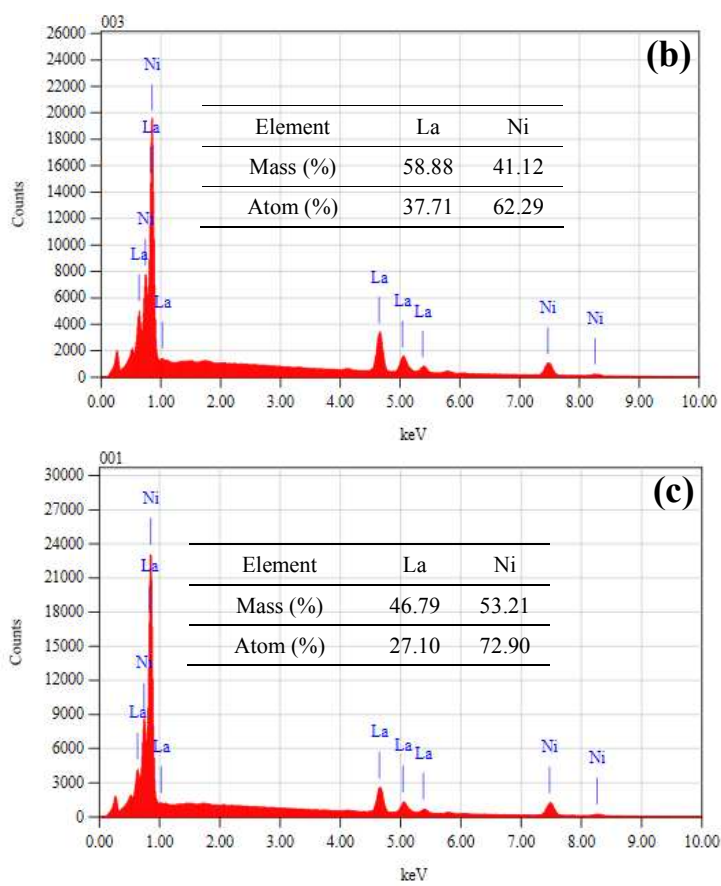
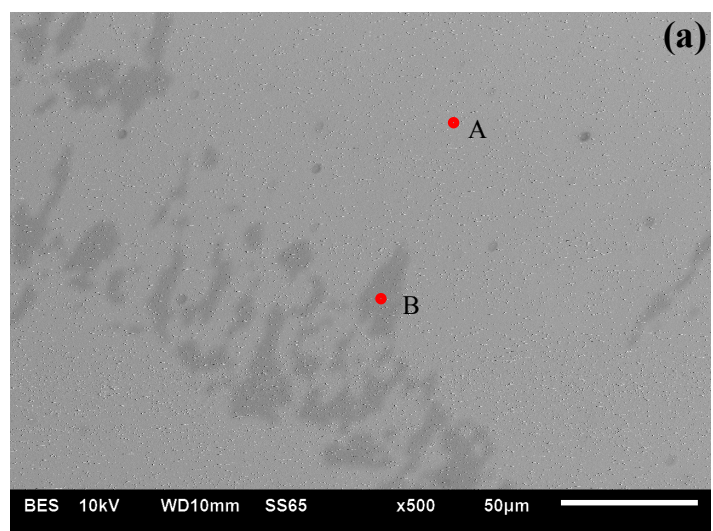


Fig.2



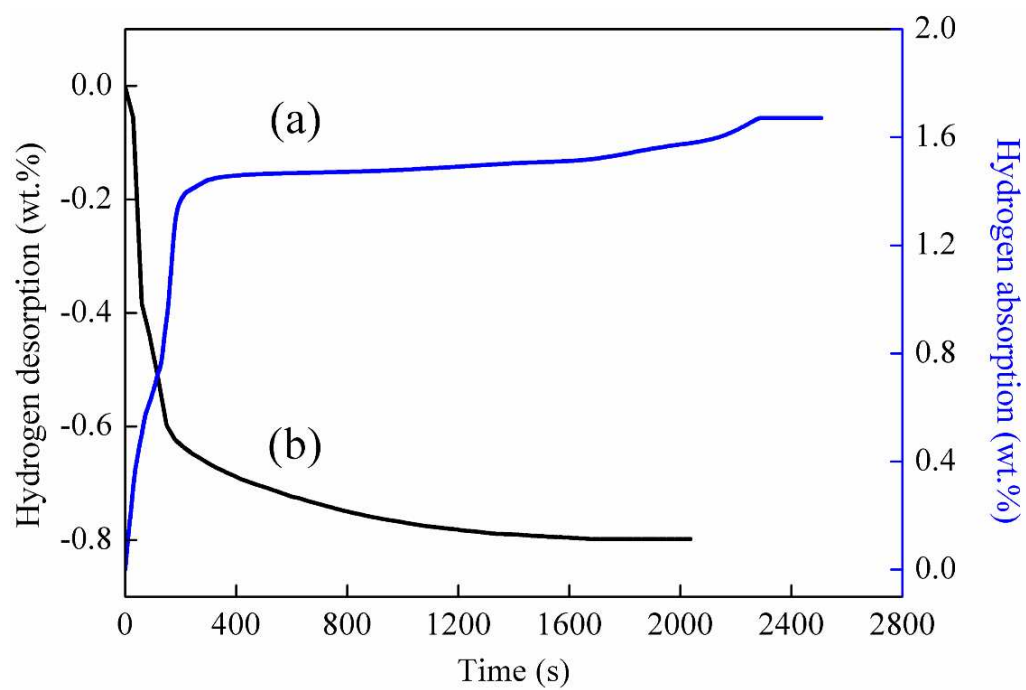


Fig. 3

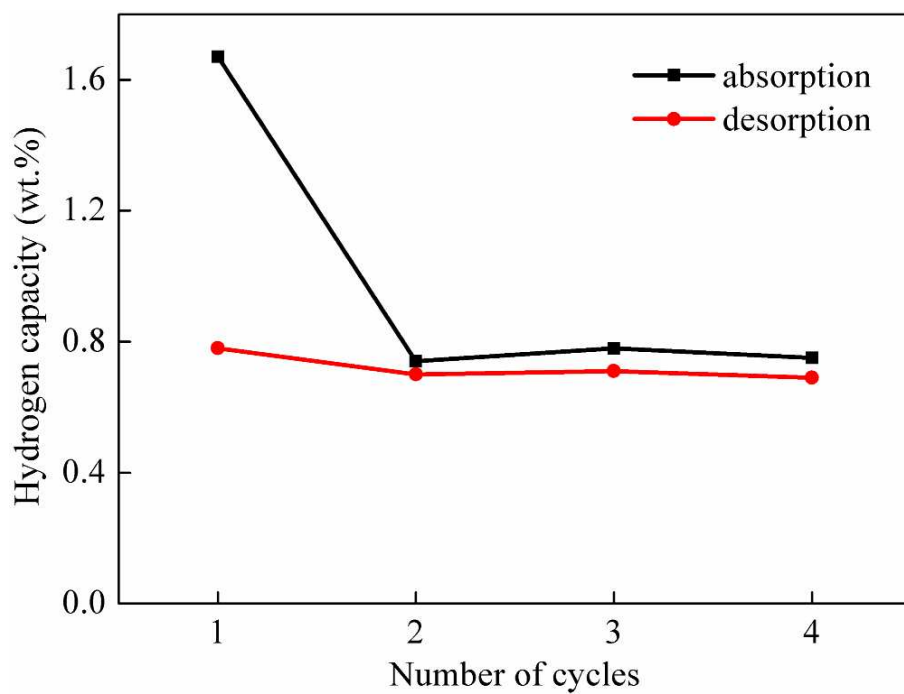
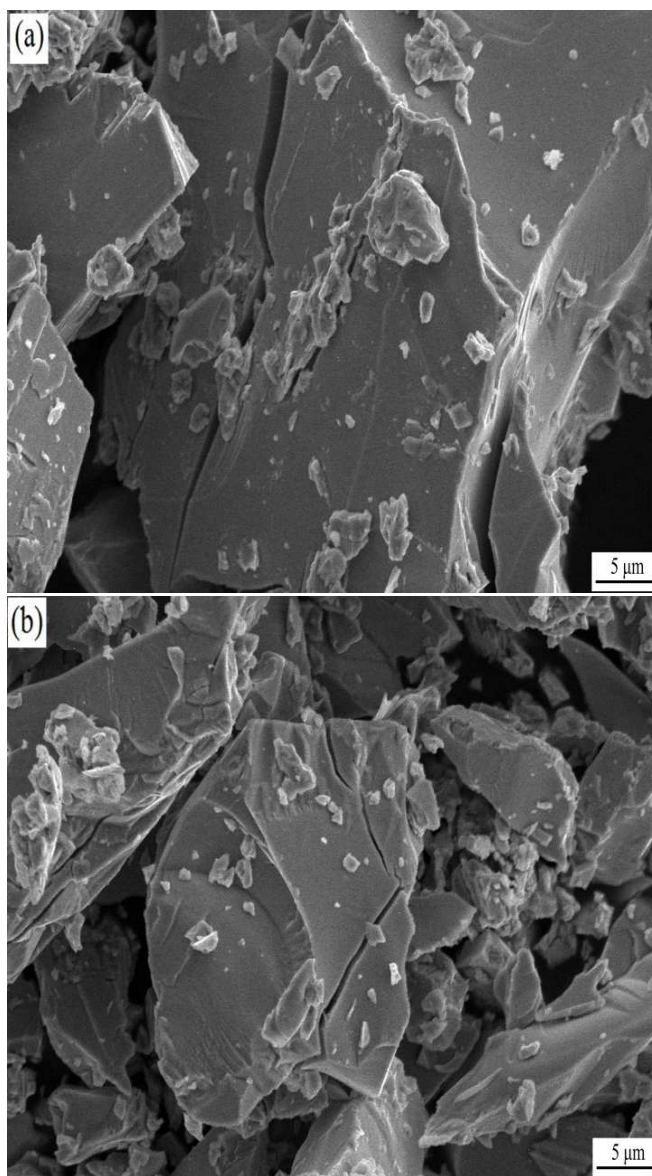
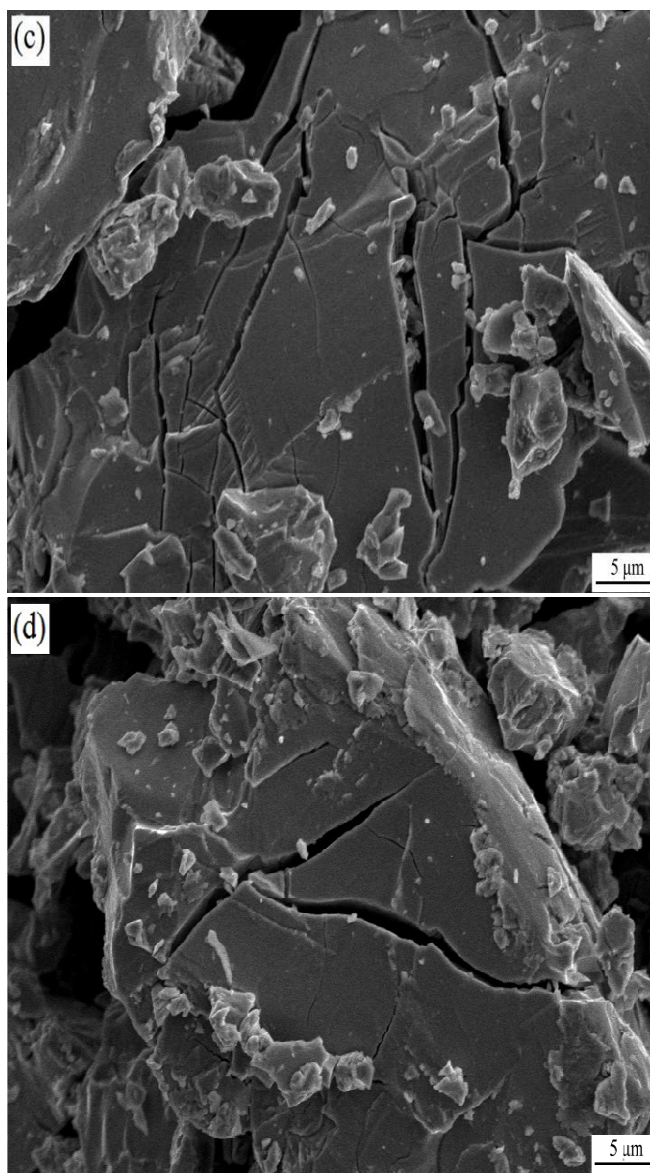


Fig. 4





**Fig. 5**

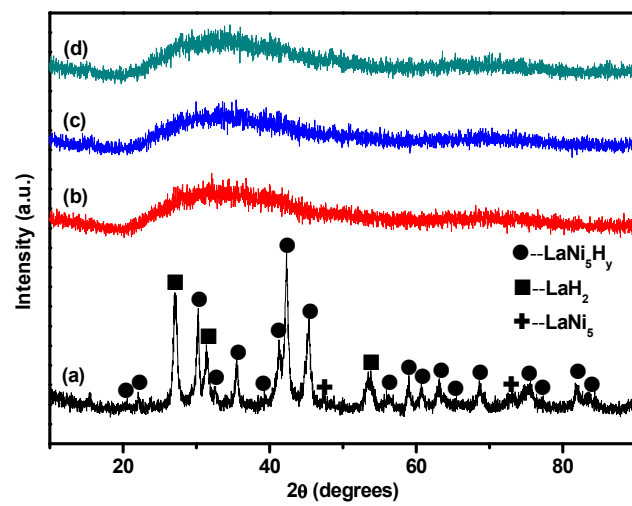
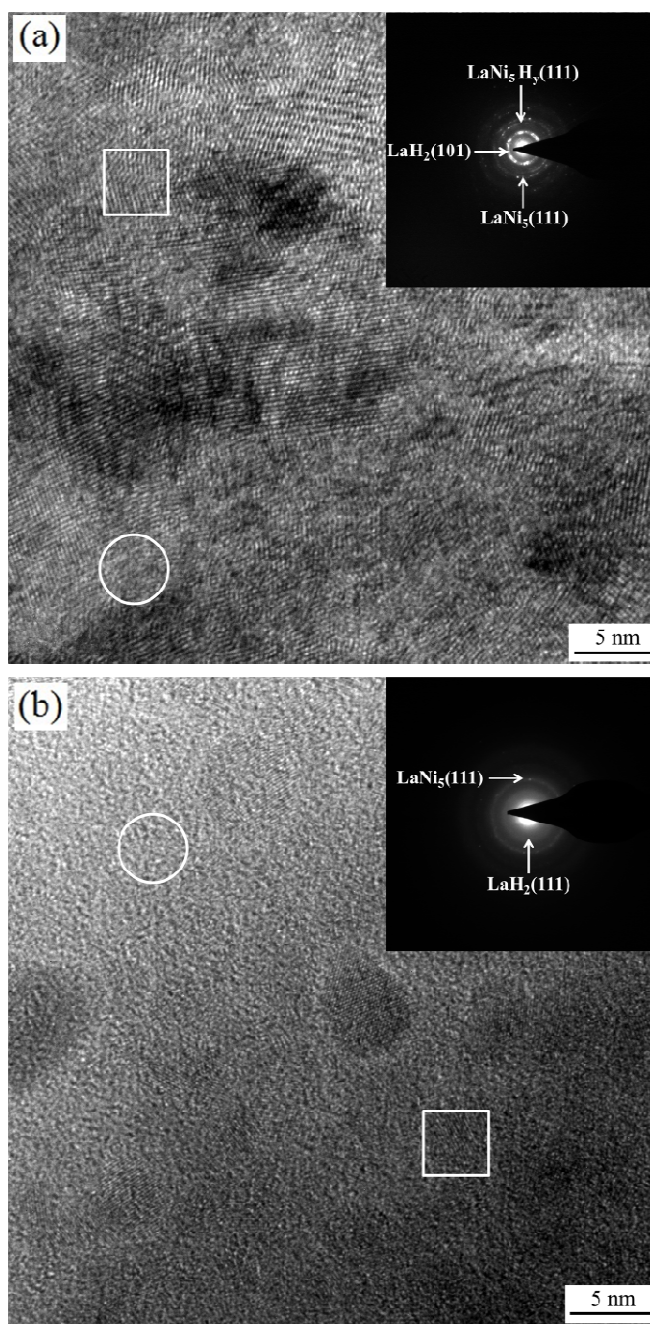


Fig. 6



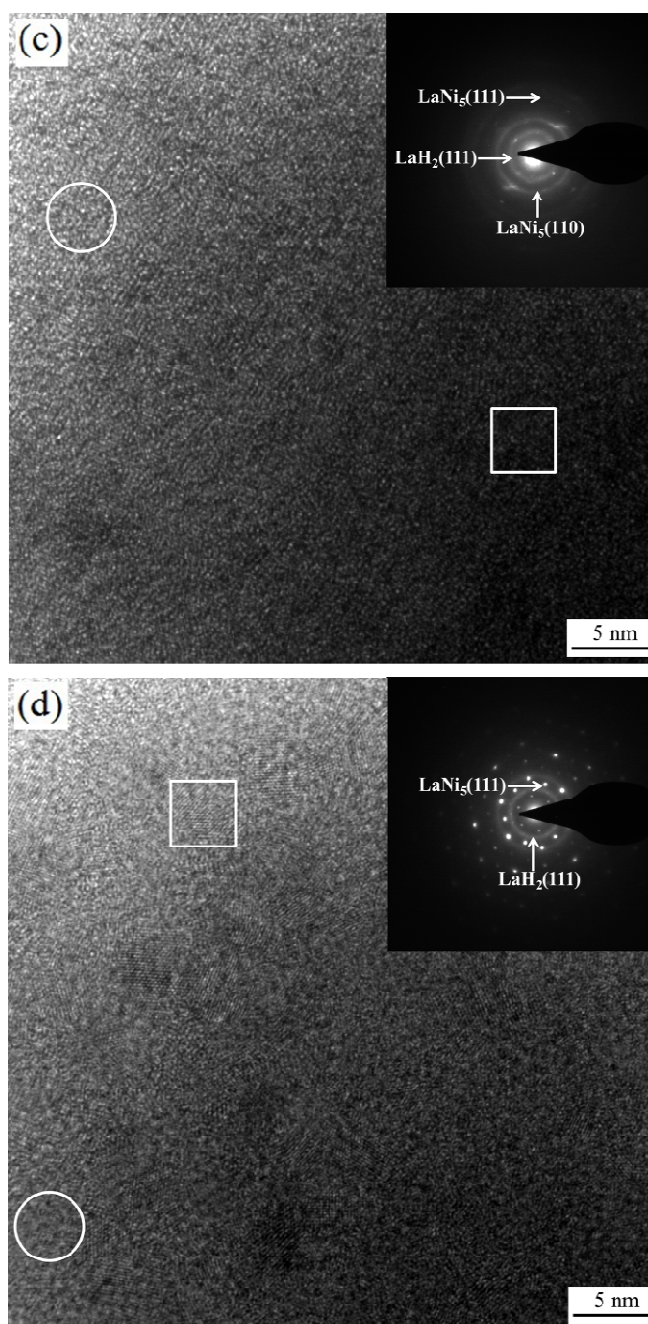


Fig. 7