



## Study of milling time impact on hydrogen desorption from $\text{LiAlH}_4\text{-Fe}_2\text{O}_3$ composites

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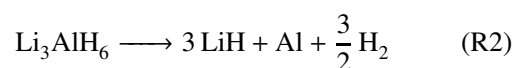
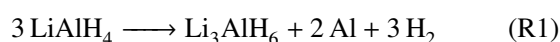
### Abstract

$\text{LiAlH}_4$  was modified by mechanical milling and with the addition of 5 wt.%  $\text{Fe}_2\text{O}_3$  in order to improve its hydrogen desorption properties. The composite was milled for 1, 3, 5, 7 or 15 min, and depending on the milling time, various phenomena took place. Up to a milling time of 5 min, the particle size of the composite decreases. Further milling leads to the particles agglomeration reaching the size of the starting material after 15 min. Moreover, the mechanical milling process leads to the transformation of  $\text{AlH}_4^-$  to  $\text{AlH}_6^{3-}$  structure as a result of partial hydrogen desorption. Hydrogen desorption during the milling is the most pronounced in the sample milled for 15 min, so this sample has only one hydrogen desorption peak in the temperature-programmed desorption measurements. Mechanical milling with the addition of  $\text{Fe}_2\text{O}_3$  for up to 15 min improves  $\text{LiAlH}_4$  hydrogen desorption properties as hydrogen desorption temperature and apparent activation energies decrease.

**Keywords:** hydrides, composites, mechanochemical synthesis, hydrogen storage, kinetics

### 1. Introduction

Metal and complex light hydrides are the best-fitted materials for hydrogen storage within the concept of hydrogen-based economy [1,2]. These materials meet the basic application requirements: low-cost, safe usage (stable compounds and composites based on them), and they are environmentally friendly (reversible materials with a few thousands of sorption cycles lifespan). However, as relatively stable compounds they exhibit some undesirable properties such as sluggish dehydrogenation kinetics and high hydrogen desorption temperature. Among various hydrides,  $\text{LiAlH}_4$  (space group  $P2_1/c$ ) emerged as an attractive candidate for solid-state hydrogen storage at moderate pressures and temperatures [3]. This compound decomposes in three steps, according to the following reactions [3,4]:



The first reaction occurs in the temperature range of 150–175 °C, the second between 180–220 °C and the third one between 400–420 °C. The first two reactions are significant from the hydrogen storage point of view: i) both take place at a reasonably low temperature (R2 ends at 220 °C) and ii) the overall sum of their gravimetric hydrogen capacity is 7.9 wt.%. Thus, the reactions R1 and R2 are accessible for practical hydrogen storage. However, slow hydrogen desorption kinetics of the pure hydride and irreversibility of cycling are still bottlenecks for the on-board application of lithium-alanate.

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Nevertheless,  $\text{LiAlH}_4$  is suitable for synthesizing the composites with 5 or 10 wt.% of additive. The additive should improve the dehydrogenation process of hydride without a significant decrease in storage capacity, so these composites could still meet the US DoE's target of 5.5 wt.%  $\text{H}_2$  storage capacity for practical application [5].

Various additives like pure metals [6–8], carbides [9,10], halides [11–15], carbon based materials [16], metallic oxides and oxide ceramics [17–24] and hydrides [12,25] were used in order to improve the hydrogen sorption properties of  $\text{LiAlH}_4$ . These improvements are achieved due to the reduction of the particle size of hydrides to the nanoscale, an increase in reactive particle surface area, destabilization of the hydride crystal structure and catalytic performance of additives. Li *et al.* [17] showed that the onset dehydrogenation temperature is about 80 °C lower for the reaction R1 and 45–60 °C lower for the second stage (R2) when 5 mol% of  $\text{Fe}_2\text{O}_3$  and  $\text{Co}_2\text{O}_3$  are added to  $\text{LiAlH}_4$ . The addition of these oxides also improves the reaction kinetics. The composites release 7.1 wt.% of hydrogen in 70 min (at 120 °C), compared to the as-received  $\text{LiAlH}_4$  that desorbs 0.3 wt.% of  $\text{H}_2$  under the same conditions. So, the apparent activation energies ( $E_{\text{app}}$ ) for both desorption steps are considerably lower than in the pure  $\text{LiAlH}_4$ . Doping with Fe- $\text{Fe}_2\text{O}_3$  promotes formation of Fe-Al and Fe- $\text{Al}_2\text{O}_3$  intermetallics during dehydrogenation of  $\text{LiAlH}_4$ . These phases have beneficial impact on the hydrogen desorption kinetics by promoting diffusion of hydrogen ions [22]. Ismail *et al.* [21] have demonstrated that adding a small amount of  $\text{TiO}_2$  to  $\text{LiAlH}_4$  results in a significant 90 °C reduction in the decomposition temperature compared to the as-received  $\text{LiAlH}_4$ . The composite material starts releasing hydrogen at 60 °C, so dehydrogenation is completed below 200 °C, with approximately 7.5 wt.% of  $\text{H}_2$  desorbed. Changing  $\text{TiO}_2$  to Sr-titanate also decreases both the hydrogen release temperature and activation energies [23]. When added 7 wt.% of  $\text{NiTiO}_3$  @h-BN to  $\text{LiAlH}_4$  the R1 starts at 68 °C, and activation energies for both R1 and R2 are significantly reduced. The absorption of hydrogen at 30 bar and 300 °C achieved approximately 1 wt.% [24]. In  $\text{LiAlH}_4$  composite with 10 wt.% of  $\text{K}_2\text{NbF}_7$  activation energies decreased for 24 (R1) and 26 kJ/mol (R2) compared to the as-milled  $\text{LiAlH}_4$ . Improvement in the desorption kinetics performance was attributed to the *in situ* formation of  $\text{NbF}_4$ , LiF and K or K-containing phases that appeared during the heating pro-

cess [11]. On the other hand in  $\text{LiAlH}_4$  doped with  $\text{K}_2\text{NiFe}_6$ , AlNi and LiF phases formed during dehydrogenation and improved its kinetics [15]. Similarly, by milling  $\text{LiAlH}_4$  with 7 wt.% of  $\text{NiFe}_2\text{O}_4$  [18], hydrogen desorption energies were reduced by approximately 60%. Desorption energy barrier decreased because *in situ* formed  $\text{Al}_4\text{Ni}_3$  accelerates the breakdown of Al-H bonding through the interfacial charge transfer and the dehybridization of Al-H cluster. It supports our previous findings that the charge transfer from  $\text{LiAlH}_4$  and  $\text{Li}_3\text{AlH}_6$  to  $\text{Fe}_2\text{O}_3$  leads to weaker bonding of hydrogen atoms and improved hydrogen desorption performance of  $\text{LiAlH}_4$  [26].

High-energy ball milling proved to be the best method for the composite synthesis and additives and defects introduction in the  $\text{LiAlH}_4$  structure. However, composites often desorb less hydrogen than expected, especially under prolonged milling conditions [27]. During the milling process, the temperature in the milling chamber can significantly increase, reaching the temperature of R1 or even R2 leading to the degradation of hydride and the decrease in the hydrogen storage capacity of the material. Prevention of the hydride decomposition during the preparation was done in cryogenic ball-milling [12,13]. But this process requires the liquid nitrogen temperature. So, in this work, we studied the impact of the milling time on the hydride structure, possible degradation, and desorption properties in lithium alanate/iron(III)-oxide composites (5 wt.%).

## II. Experimental

The  $\text{LiAlH}_4$  + 5 wt.%  $\text{Fe}_2\text{O}_3$  (labelled as LiFe) composites were synthesized by mechanochemical milling in high-energy ball mill SPEX 5100 using hardened steel ball and the ball to powder ratio (BPR) 10:1. The mass of the ball was 1 g, so 100 mg of composite was synthesized after one process. Used chemicals are Alfa Aesar  $\text{LiAlH}_4$  (purity 97%) and catalyst grade Sigma Aldrich  $\text{Fe}_2\text{O}_3$ . The samples were milled for 1, 3, 5, 7 and 15 min. The sample labels are given in the Table 1. Handling of the samples and milling processes were done in the argon atmosphere.

Structural and phase characterization were done using Ultima IV Rigaku X-ray diffractometer. The diffractometer was equipped with  $\text{Cu K}_{\alpha 1,2}$  radiation source, with voltage of 40.0 kV and a current of 40.0 mA. Diffractograms were recorded in  $2\theta$  range from 10 to 80°, with recording step of 0.02° and scanning rate of

Table 1. The sample labels

Sample mark	$\text{LiAlH}_4$ [wt.%]	$\text{Fe}_2\text{O}_3$ [wt.%]	Milling time [min]
$\text{LiAlH}_4$ as-received	100	0	0
LiFe1	95	5	1
LiFe3	95	5	3
LiFe5	95	5	5
LiFe7	95	5	7
LiFe15	95	5	15

2°/min. The morphology of materials surface and their microstructure were characterized by scanning electron microscope SEM, JEOL JSM 6460LV equipped with EDS INCA microanalysis. A quantitative particle size distribution was obtained by Malvern 2000 SM MasterSizer laser scattering particle size analysis system. The specified resolution range of the system was sub- $\mu\text{m}$  to 2 mm. 2-propanol was used as a suspension media. To enhance the dispersion, all samples were ultrasonicated for 5 min prior to measurements. The same stirring speed and obscuration level were used for all samples. Fourier transform infrared spectroscopy (FTIR) was used for analysis of the pure and doped  $\text{LiAlH}_4$  samples using a Perkin Elmer Spectrum Two FT-IR spectrometer using the pressed KBr pellets technique (mass ratio 1:100). Range of the recording was from 800 to  $2000\text{ cm}^{-1}$  with a spectral resolution of  $2\text{ cm}^{-1}$ . Thermal behaviour of the samples was investigated by temperature programmed desorption (TPD). A custom-built set-up for TPD was used. The composites of lithium alanes doped with iron(III)-oxide in the form of powders were put in the sample holder made of quartz. Quartz tube was placed inside an electrical furnace and coupled with an Extorr 3000 quadrupole mass spectrometer (Extorr Inc.). The signals, as partial pressures (torr) at 7 different  $m/z$  ratios were followed and simultaneously recorded as a function of time (s): 1 (H), 2 ( $\text{H}_2$ ), 17 ( $\text{NH}_3$  or OH), 18 ( $\text{H}_2\text{O}$ ), 28 (CO), 32 ( $\text{O}_2$ ), 44 ( $\text{CO}_2$ ). Up to 5 mg of the samples were placed in the quartz tube, outgassed to  $10^{-7}$  torr and then subjected to TPD at a linear heating rate of  $10^\circ\text{C}/\text{min}$ , from room temperature to  $350^\circ\text{C}$ .

### III. Results and discussion

#### 3.1. Microstructural characterization

X-ray diffraction patterns of the as-received (unmilled)  $\text{LiAlH}_4$ ,  $\text{Fe}_2\text{O}_3$  and their composites  $\text{LiAlH}_4$ -5 wt.%  $\text{Fe}_2\text{O}_3$  milled for 1, 3, 5, 7 and 15 min are shown in Fig. 1. All characteristic  $\text{LiAlH}_4$  peaks are observed in the  $2\theta$  range from 20 to  $50^\circ$  and correspond to monoclinic unit cell ( $P2_1/c$ ) [28]. The as-received hydride has also small amount of Al phase impurity, whose characteristic maxima are at  $38.5$  and  $44.5^\circ$  [27]. There are no visible peaks from oxide or hydroxide impurities [29,30]. The composite samples show diffraction maxima originating from  $\text{LiAlH}_4$  and  $\text{Fe}_2\text{O}_3$  phase. As milling time increases the intensity of character-

istic lithium-alanate phase maxima decreases and becomes broader. After 15 min of high energy milling, there are no visible  $\text{LiAlH}_4$  phase peaks. On the other hand, intensity of aluminium peaks increases, indicating the hydride decomposition. There is also broad peak in the sample milled for 15 min at approximately  $22^\circ$  that could correspond to  $\text{Li}_3\text{AlH}_6$  phase maxima at  $21.95^\circ$  and  $22.55^\circ$  [27]. The structure of  $\text{LiAlH}_4$  is prone to destabilization by mechanochemical treatment, so this phase transformation happens if desorption of  $\text{H}_2$  (reaction R1) takes place during milling [27,28]. During this transformation lithium hexahydridoaluminate ( $\text{Li}_3\text{AlH}_6$ ) and Al phases are gradually generated. The characteristic  $\text{Fe}_2\text{O}_3$  peaks also become more pronounced with the milling time increase [17].

The reduction in hydride crystallite size in the composite upon milling for various times is calculated by Debye-Scherrer equation from the broadening of 101 peak and the results are presented in Table 2 [31]. It can be seen that 3 min of milling leads to the decrease in crystallite size for 32%, while further milling does not have any effect. The shape of the unmilled  $\text{LiAlH}_4$  particles is relatively regular with sharp edges that resemble the monoclinic crystal lattice of hydride (Fig. 2a). The particle size distribution is uniform. The ball milling impact on the hydride particle morphology is visible in Figs. 2c,d for the composite milled for 5 min. Milling

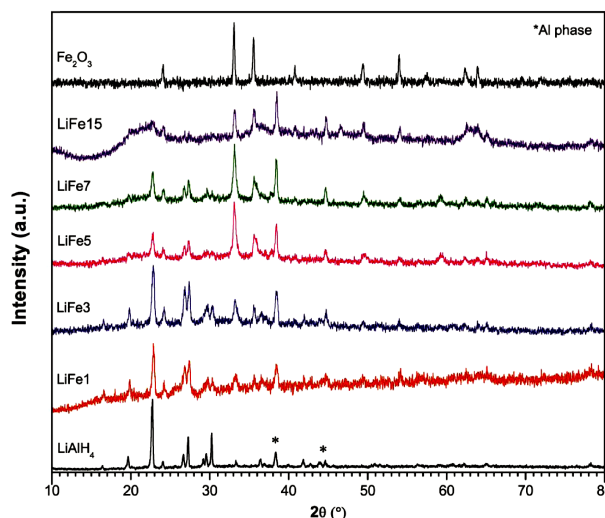
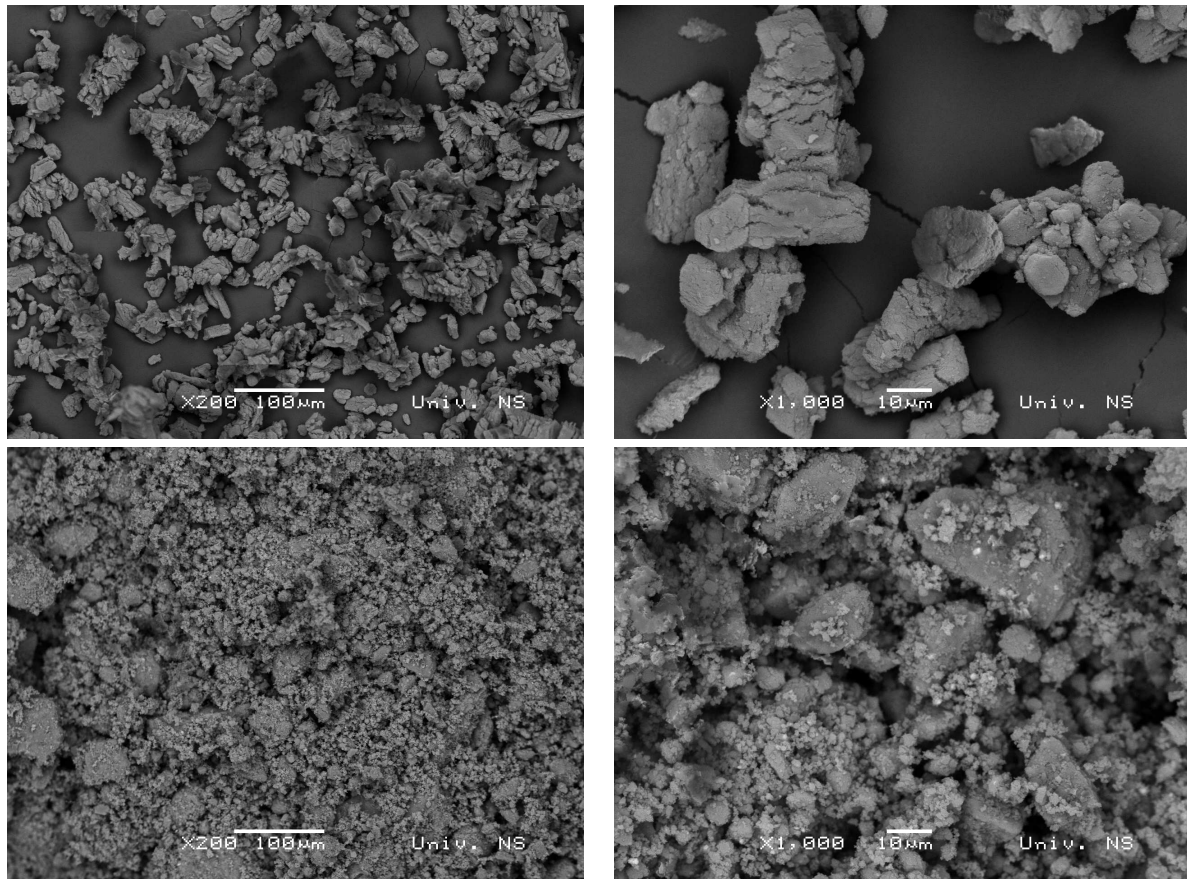


Figure 1. XRD patterns of unmilled  $\text{LiAlH}_4$  and  $\text{Fe}_2\text{O}_3$  phases and  $\text{LiAlH}_4$ - $\text{Fe}_2\text{O}_3$  composites milled for 1 (LiFe1), 3 (LiFe3), 5 (LiFe5), 7 (LiFe7) and 15 min (LiFe15)

Table 2. Interplanar distance ( $d$ ), crystallite size ( $D$ ) and mean particle size of as-received  $\text{LiAlH}_4$  and composite samples

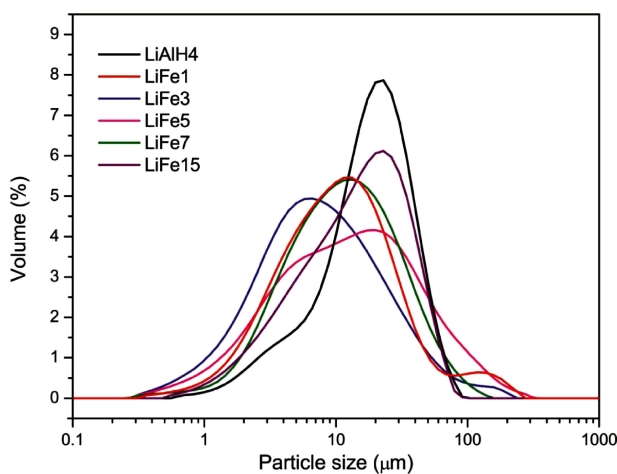
Sample	$2\theta$ [ $^\circ$ ]	$d$ [ $\text{\AA}$ ]	$D$ [nm]	Mean particle size [ $\mu\text{m}$ ] (left/right peak)
$\text{LiAlH}_4$	22.68	3.917	33	22
LiFe1	22.74	3.907	25	12/130
LiFe3	22.79	3.899	22	6/145
LiFe5	22.77	3.902	24	5/20
LiFe7	22.79	3.899	24	12
LiFe15	22.79	3.899	7	23



**Figure 2.** SEM micrographs of unmilled  $\text{LiAlH}_4$  (a and b) and  $\text{LiAlH}_4\text{-Fe}_2\text{O}_3$  composite milled for 5 min - LiFe5 (c and d)

led to the significant refinement of particles. Milled particles have irregular shape with sponge-like structure and broad particle size distribution. There is also visible agglomeration of particles in the milled sample.

The results of the particle size distribution obtained by laser scattering (LS) measurements are presented in Fig. 3. In the unmilled  $\text{LiAlH}_4$  the 99% of sample volume consists of particles in the range between 1 and 100 μm with the mean particle size of 22 μm. Distri-



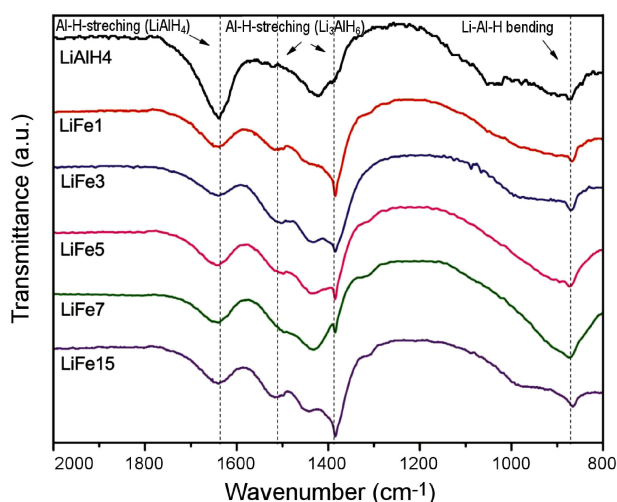
**Figure 3.** Particle size distribution (PSD) of unmilled  $\text{LiAlH}_4$  and  $\text{LiAlH}_4\text{-Fe}_2\text{O}_3$  composites milled for 1 (LiFe1), 3 (LiFe3), 5 (LiFe5), 7 (LiFe7), and 15 min (LiFe15)

bution of particles is not fully monomodal, but there is a shoulder in the range between 1 and 10 μm with the maximum at 3 μm. The sample LiFe1 shows bimodal distribution of particles in the range from 300 nm to 300 μm, where 73% of the volume consists of smaller particles with the mean size of 12 μm while 27% of particles are between 100 and 300 μm, with the mean particle size of 130 μm. In comparison to the unmilled  $\text{LiAlH}_4$ , main peak is shifted to smaller particles in the range from 22 to 12 μm. The composite milled for 3 min, LiFe3, shows also bimodal distribution: 86% of particle volume is in the range from 1 to 100 μm with the mean particle size of 6 μm, while 14% of volume is between 100 and 200 μm (mean particle size 145 μm). In the LiFe5 sample 0.2% of particles volume is in the range from 250 nm to 1 μm, 78% in the range from 1 to 100 μm and 21.8% between 100 μm to 300 μm. Distribution is bimodal with the two peaks maxima at 5 and 20 μm. PSD analysis of the LiFe7 shows only one peak which characterizes very high symmetry (bell-like shape). Almost all the particles are distributed in the range from 1 to 100 μm (97.2% of particles) and rest of particles are distributed in range from 100 to 160 μm (2.8% of particles). Mean particle size is 12.5 μm. The LiFe15 composite has also monomodal distribution, but the mean particle size is 23 μm. All particles are distributed in the range from 0.5 to 100 μm. The milling process leads to the decrease of particle size up to 5 min



milling, afterwards particles start to agglomerate. Up to 7 min milling there are two types of particles: bigger and smaller than 100  $\mu\text{m}$ . It is interesting that after 7 min of milling, distribution is monomodal and almost without particles bigger than 100  $\mu\text{m}$ , which is similar in the sample milled for 15 min. The mean particle size in the composite milled for 15 min is similar to the mean particle size of the starting hydride.

Fourier transform infrared (FTIR) spectra of the as-received  $\text{LiAlH}_4$  and milled  $\text{LiAlH}_4$ -5 wt.%  $\text{Fe}_2\text{O}_3$  composites are shown in Fig. 4. Whole spectrum is in the 800–2000  $\text{cm}^{-1}$  wavenumber range.  $\text{LiAlH}_4$  has two active infrared vibrations: Al–H stretching mode at 1641  $\text{cm}^{-1}$  and Li–Al–H bending mode at 885  $\text{cm}^{-1}$  [17]. As milling starts, the stretching mode becomes less pronounced with the appearance of Al–H stretching mode from hexahydridoaluminate ions ( $\text{AlH}_6^{3-}$ ) in  $\text{Li}_3\text{AlH}_6$  at 1385 and 1500  $\text{cm}^{-1}$  [14,17]. This peak increases with the milling time indicating decomposition of  $\text{LiAlH}_4$  via reaction R1 which is in accordance with XRD findings. So, the addition of  $\text{Fe}_2\text{O}_3$  facilitates the  $\text{H}_2$  release from  $\text{LiAlH}_4$  and the dehydrogenation process begins very fast when milling in high energy mill starts.



**Figure 4.** FTIR of the unmilled  $\text{LiAlH}_4$  and  $\text{LiAlH}_4$ -5 wt.%  $\text{Fe}_2\text{O}_3$  composites milled for 1 (LiFe1), 3 (LiFe3), 5 (LiFe5), 7 (LiFe7) and 15 min (LiFe15)

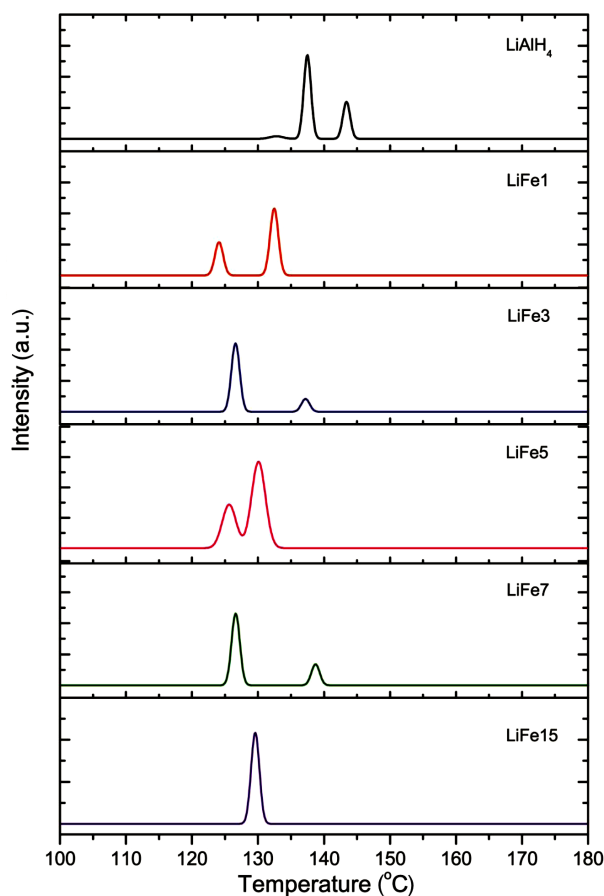
### 3.2. Analysis of dehydrogenation properties

To study the dehydrogenation properties of the pure  $\text{LiAlH}_4$  and composite materials milled for different times, TPD-MS was done. Desorption of hydrogen obtained from TPD measurements is presented in Fig. 5. As shown in Fig. 5, there are two distinguished  $\text{H}_2$  desorption maxima for the unmilled  $\text{LiAlH}_4$  at 137 (LT) and 143  $^\circ\text{C}$  (HT) corresponding to the reactions R1 and R2, respectively. There is also very small peak at 132.5  $^\circ\text{C}$  which could originate from the smaller particles (1–10  $\mu\text{m}$ ) that form shoulder in particle size distribution (Fig. 2) in this sample. The addition of 5 wt.%  $\text{Fe}_2\text{O}_3$  by mechanical milling leads to the decrease in

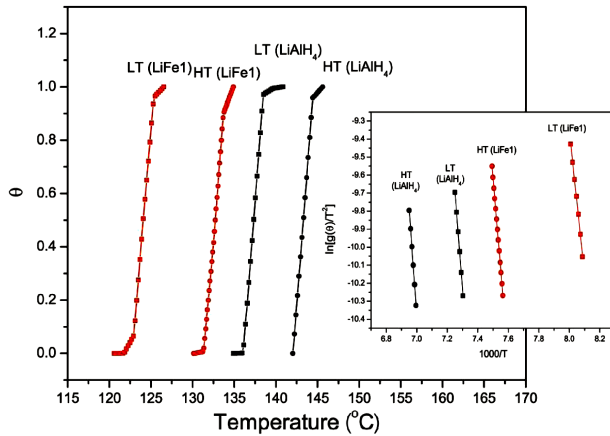
hydrogen desorption temperature and the dependence of hydrogen desorption temperature on time can be noticed. The one minute milled  $\text{LiAlH}_4$ -5 wt.%  $\text{Fe}_2\text{O}_3$  composite, LiFe1, releases hydrogen at 124 and 133  $^\circ\text{C}$ , but the first peak at lower temperature (LT) is smaller than the second one.

After 3 min of milling, the composite releases hydrogen at 126 and 137  $^\circ\text{C}$  with the higher amount of hydrogen desorbed by reaction R1. The further milling time increase, up to 15 min, does not influence the desorption temperature to a large extent, but it changes the portions of hydrogen released by reactions R1 and R2. Finally, 15 min of milling leads to the one step hydrogen desorption. As XRD and FTIR analyses showed, there is certain  $\text{H}_2$  release during the milling process by reaction R1, where the destabilized structure of  $\text{LiAlH}_4$  phase decomposes to lithium hexahydridoaluminate phase [27]. This is the most pronounced in the composite milled for 15 min, so the TPD single peak can be attributed to the hydrogen release completely by only reaction R2.

Kinetics and mechanism of the solid-state reactions can be modelled with a few models depending on the sorption reaction limiting step [32]. Šestak and Berggren [33] explored the possibility to obtain the



**Figure 5.** Temperature programmed desorption (TPD) spectra of the unmilled  $\text{LiAlH}_4$  and  $\text{LiAlH}_4$ -5 wt.%  $\text{Fe}_2\text{O}_3$  composites milled for 1 (LiFe1), 3 (LiFe3), 5 (LiFe5), 7 (LiFe7), and 15 min (LiFe15)



**Figure 6.** Temperature evolution of the reacted fraction ( $\theta$ ) corresponding to  $\text{LiAlH}_4$  hydrogen desorption, obtained by integration of TPD peaks for unmilled hydride ( $\text{LiAlH}_4$ ) and composite  $\text{LiFe1}$  milled for 1 min (insert: experimental data and the best fit obtained for nucleation and growth model  $g(\theta) = [-\ln(1 - \theta)]^{1/2}$ )

mechanism of thermal decomposition process from the non-isothermal curves. Detailed application of this method on the reaction of hydrogen desorption from hydrides is given in our previous work [32]. Desorption curves of the  $\text{LiAlH}_4$  and its composites with  $\text{Fe}_2\text{O}_3$ , have been analysed using different kinetic models. The analysis of TPD curves indicates that the  $\text{H}_2$  desorption from the  $\text{LiAlH}_4$  and its composites with  $\text{Fe}_2\text{O}_3$  are controlled by nucleation and growth mechanisms with the Avrami parameter  $n = 2$  over a coverage  $\theta$  range from 0.3 to 0.8 (Fig. 6).

Parameter  $n = 2$  corresponds to two-dimensional (2D) nuclei growth. In this model, the reacted fraction  $\theta$

can be related to temperature ( $T$ ) by:

$$\ln \left[ \frac{[-\ln(1 - \theta)]^{1/n}}{T^2} \right] = f \left( \frac{1000}{T} \right) \quad (1)$$

Calculated apparent activation energies  $E_{app}$  (kJ/mol) for all samples and both hydrogen desorption steps (reactions R1 and R2) are given in Table 3.

It can be seen that mechanical milling of the composite with  $\text{Fe}_2\text{O}_3$  even as short as one minute significantly improves the kinetics of reactions R1 and R2. The lowest desorption temperatures and apparent activation energies  $E_{app}$  are in the composite sample milled for five minutes. The values of apparent activation energies are comparable to the ones found in the literature obtained by the Kissinger method (Table 4). In our previous work, we have discussed the role of  $\text{Fe}_2\text{O}_3$  on hydrogen desorption from  $\text{LiAlH}_4$  [26]. We showed that a significant portion of the iron ions has changed their valence state from  $\text{Fe}^{3+}$  to the  $\text{Fe}^0$  and  $\text{Fe}^{2+}$  after hydride decomposition, so there is a transfer of an electron to  $\text{Fe}_2\text{O}_3$  during hydrogen desorption. The effect of charge transfer from  $\text{LiAlH}_4$  and  $\text{Li}_3\text{AlH}_6$  to  $\text{Fe}_2\text{O}_3$  leads to weaker bonding of hydrogen atoms and easier hydrogen desorption.

#### IV. Conclusions

The presented results lead to conclusion that the composite of  $\text{LiAlH}_4$  with  $\text{Fe}_2\text{O}_3$  milled for as short as 1 min in a high energy mill shows significantly better desorption properties than the unmilled  $\text{LiAlH}_4$ , i.e. i) temperature peaks of hydrogen desorption are shifted to lower positions for all composites; ii) increase of milling time

**Table 3.**  $\text{H}_2$  desorption temperature and apparent activation energy obtained from TPD measurements (LT - lower temperature peak, HT - higher temperature peak)

Sample	Temperature (LT) [°C]	$E_{app,LT}$ [kJ/mol]	Temperature (HT) [°C]	$E_{app,HT}$ [kJ/mol]
$\text{LiAlH}_4$	137	$91 \pm 1$	143	$98 \pm 1$
$\text{LiFe1}$	124	$66 \pm 2$	133	$83.2 \pm 0.3$
$\text{LiFe3}$	126	$77 \pm 1$	137	$88 \pm 0.5$
$\text{LiFe5}$	125	$55 \pm 2$	130	$69 \pm 3$
$\text{LiFe7}$	127	$72 \pm 1$	139	$89 \pm 1$
$\text{LiFe15}$	129	$70 \pm 3$	-	-

**Table 4.** Apparent activation energy ( $E_{app}$ ) of as-received  $\text{LiAlH}_4$  and  $\text{LiAlH}_4$  doped with various catalysts, calculated by the Kissinger method

$E_{app}$ [kJ/mol] (LT/HT)	Compound	Experimental conditions	Reference
81/108	$\text{LiAlH}_4$	As received	[8]
89/103	$\text{LiAlH}_4$ - 2 mol% $\text{TiCl}_3$ *1/3 $\text{AlCl}_3$	Milled for 1 min	
100/130	$\text{LiAlH}_4$	Milled for 90 min	[27]
80/100	$\text{LiAlH}_4$ - 5 wt.% 2 $\text{DTi}_2\text{C}_3$	Milled for 10 h	
111/100	$\text{LiAlH}_4$	As received	[34]
92.5/92	$\text{LiAlH}_4$	Milled for 15 min	
102/110	$\text{LiAlH}_4$	As received	[20]
84/96	$\text{LiAlH}_4$ - 5 wt.% $\text{Fe}_2\text{O}_3$	Milled for 60 min	
95/172	$\text{LiAlH}_4$	As received	[17]
54/86	$\text{LiAlH}_4$ - 5 mol% $\text{Fe}_2\text{O}_3$	Milled for 30 min	
56/93	$\text{LiAlH}_4$ - 5 mol% $\text{Cr}_2\text{O}_3$	Milled for 30 min	

causes the change in the intensity of two characteristic hydrogen desorption peaks; iii) the activation energy for all milled composites is lower in comparison to the unmilled  $\text{LiAlH}_4$ . During the process of high energy mechanical milling, hydrogen desorption by R1 reaction starts, so the composites milled for 15 min have only one desorption maxima corresponding to R2 reaction. Using the ball milling approach to synthesize the nanostructured hydrides can significantly enhance the thermodynamics and kinetics of dehydrogenation. However, there is necessary to apply the minimal milling time. This minimal milling time should be long enough to assure the best composite properties, but not to start the hydrogen desorption. For explored milling conditions (experimental part) optimal results are obtained in the sample milled for 5 min. It is assumed that this milling time is long enough to disperse the catalyst homogeneously around  $\text{LiAlH}_4$  particles and also cause the maximal reduction in the particle size. Further milling leads to the particle agglomeration, but more importantly to the significant hydrogen release during milling.

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