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Microstructure and hydrogen storage properties of as-cast and rapidly solidified Ti-rich Ti-V alloys

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Abstract: The goal of the present work was to optimize the phase-structural composition and microstructure of binary $Ti_{0.8-0.9}V_{0.2-0.1}$ alloys with respect to their hydrogen sorption properties. Application of these alloys is for hydrogen absorption from gaseous mixtures containing substantial amounts of carbon monoxide (CO) at high temperatures. Irrespective of alloy composition, both α (HCP) and β (BCC) phases in $Ti_{0.8-0.9}V_{0.2-0.1}$ formed single phase FCC hydrides upon hydrogenation in pure H₂. An in situ synchrotron X-ray diffraction study showed that only the β -phase transformed to the corresponding hydride when the alloy was hydrogenated in a mixture of H₂+10%CO. Rapid solidification (RS) of the alloy resulted in refined grain sizes both in the $Ti_{0.8}V_{0.2}$ and $Ti_{0.9}V_{0.1}$ alloys. Furthermore, RS was found to increase the β -phase fraction in $Ti_{0.9}V_{0.1}$, being twice larger than that of the as-cast alloy. $Ti_{0.9}V_{0.1}$ had a platelike microstructure as observed by scanning electron microscopy (SEM), the plates were about 300 nm thick. The microstructure refinement resulted in a faster kinetics of H desorption as observed by temperature desorption spectroscopy (TDS).

Key words: hydrogen storage; Ti-V alloys; rapid solidification; synchrotron X-ray diffraction; carbon monoxide

1 Introduction

The hydrogen storage capacity of TiH₂ and VH₂ is high, up to 4% H (mass fraction) or 150 kg/m³. The volumetric density of hydrogen is twice superior to that of the liquid hydrogen. Alloys of Ti and V form dihydrides which have hydride formation enthalpies between those of the elemental binary hydrides, i.e. between -40 to -130 kJ/mol H₂. Ti and V between 2.7% and 80% (mole fraction) vanadium form solid solution alloys [1]. The lattice parameters of these Ti_{1-y}V_y alloys are known to be dependent on the composition. Due to larger atomic radius of Ti than that of V, they decrease upon increasing the V content [2,3]. The most common production method of the BCC Ti-based alloy is conventional arc melting. Another production method, such as high energy ball milling of two elements, has also been employed to produce the alloys [4–6]. The ascast $Ti_{1-y}V_y$ alloys can be hydrogenated after activation treatments and form FCC dihydrides in the fully hydrogen-saturated condition.

Rapid solidification is one of the production methods which allow a controllable alloy solidification rate; the latter determines the microstructure and crystal structure. This method has been extensively applied for production of amorphous metals for structural or electronic applications. Recently, it has also been applied for preparation of the alloys for hydrogen storage, characterized by micro- or nano-grained microstructures. For example, promising results of using the rapid solidification technique have allowed production of nanostructured Mg–Mm–Ni composites [7,8] and Ti–V based alloys [9,10]. Some investigators showed that

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rapid solidification improved cycling properties of the alloy hydrides [11,12].

The goal of the present work is to optimize the phase-structural composition and microstructure of binary Ti–V alloys with 20% and 10% V with respect to their hydrogen sorption properties. The optimized alloy is planned to be used in a hydrogen production process, i.e. in a hydrogen sorption enhanced reactor (HSER) of steam reforming or water-gas shift processes [13]. In such a reactor, the hydrogen is absorbed in-situ from the steam reformate gas mixture, thus increasing the reactor efficiency. Since the reformate gases contain detrimental gases for hydrogen absorbing alloys, such as CO and H_2O , electroless deposition of Pd on the metal hydride alloy to increase resistance towards such active gases is also tested out.

2 Experimental

Alloys of $Ti_{0.8}V_{0.2}$ and $Ti_{0.9}V_{0.1}$ were synthesized by argon arc melting. About 5 g samples were produced during each melting. The alloy buttons were cut into smaller pieces for hydrogenation in their as-cast condition, and some of them were used as pre-alloys for rapid solidification using the melt spinning technique. Melt spinning is a rapid solidification technique, in which liquid metal is ejected through nozzle onto a rotating wheel (spinner) by which thin metal ribbons are produced. By varying the rotation speed of the spinner, the solidification rate can be controlled. This results in distinct morphology and microstructure of the metallic ribbons. In this work, a copper wheel with diameter of 30 cm and a rotation velocity of 2000 r/min were used, giving a surface velocity of about 21 m/s.

Microstructural characterization and quantification of the alloys and hydrides were performed using a Leica MeF4 light optical microscope, Zeiss Ultra 55LE field emission scanning electron microscope (FE-SEM) and JEOL JXA-8500F electron probe micro-analyzer (EPMA). The SEM and EPMA samples were prepared from both as-cast alloys and ribbons. The samples were hot mounted with conductive resin, ground with abrasive paper down to 2400 grid and mechanically polished with 90% oxide polishing suspension (OPS) +10% H₂O₂. During the SEM investigation, images were taken in backscattered mode at 20 kV and 120 µm objective aperture size. The hydrogenation properties were studied by using the volumetric thermal desorption spectroscopy (TDS) technique. The saturated dihydrides were used in the dehydrogenation experiments. The sample mass was approximately 1 g. TDS was performed by heating the samples in vacuum at a heating rate of 5 °C /min; the temperature range was from 20 °C to 800 °C.

Ex situ and in situ synchrotron diffraction (SR-XRD)

studies were applied to probe the phase-structural composition and to yield the crystallographic data of the alloys and their hydrides, which were conducted at the Swiss-Norwegian Beam Lines (SNBL), Grenoble, France. Samples with an electroless deposited Pd metal with a particle size between 40 and 60 nm were used. The method used for Pd deposition has been described in previous publications [14-16]. Small amounts of hydride powders were inserted into a quartz capillary (0.5 mm inner diameter, 0.01 mm wall thickness) and attached to a specially designed setup that was connected to a goniometer head. The in situ SR-XRD experiments were performed in an H₂ or H₂+10%CO gaseous flows of 20 mL/min (1.1 bar). The data were collected using a MAR2300 image plate detector. The wavelength was 0.6540 Å, and the sample to detector distance was 200 mm. GSAS software was used in the powder profile Rietveld refinements [7,18].

3 Results and discussion

3.1 Microstructure of as-cast and RS alloys

Figure 1 shows the microstructures of the as-cast alloys. Both $Ti_{0.8}V_{0.2}$ (Fig. 1(a)) and $Ti_{0.9}V_{0.1}$ (Fig. 1(b)) were composed of large grains with size exceeding 500 μ m. The grain size of $Ti_{0.8}V_{0.2}$ was smaller than that of $Ti_{0.9}V_{0.1}$. The composition mapping of these as-cast alloys using EMPA indicated that the alloys were



Fig. 1 Microstructures of as-cast alloys: (a) $Ti_{0.8}V_{0.2};$ (b) $Ti_{0.9}V_{0.1}$

macroscopically homogeneous. The composition of $Ti_{0.8}V_{0.2}$ measured by EMPA was 79%±2% Ti (mole fraction) and 20%±0.5% V while that of $Ti_{0.9}V_{0.1}$ was 89%±2% Ti and 10%±0.5% V.

The thickness of the ribbons produced using the melt spinning technique depended on the chemical composition of the alloys as well as the spinner velocity during preparation. By using 2000 r/m spinner velocity, ribbons with thickness of (32 ± 3) µm were produced for $Ti_{0.8}V_{0.2}$, while thicker ribbons, (49±5) µm, were formed for $Ti_{0.9}V_{0.1}$. The thickness of the ribbons correlates with the solidification rate [19]. Thinner ribbons are in general expected to be produced at higher cooling rates as compared to the thicker ribbons. However, since the alloys in this work have different compositions, direct comparison of the data is not possible. Nevertheless, it can be seen from Fig. 2(a) that RS $Ti_{0.8}V_{0.2}$ had a fine-grained microstructure with an average grain size of about 12 µm (Fig. 2(b)). This differs considerably from the RS Ti_{0.9}V_{0.1} alloy, as shown in Fig. 2(c). The microstructure of the RS Ti_{0.9}V_{0.1} resembles a typical martensitic type microstructure. A high magnification backscattered image of RS Ti_{0.9}V_{0.1} microstructure shows a plate-like microstructure with a plate thickness of about 300 nm (Fig. 2(d)).

Element mappings of RS $Ti_{0.8}V_{0.2}$ and $Ti_{0.9}V_{0.1}$ alloys examined by EPMA are shown in Fig. 3. It can be seen that both alloys are macroscopically homogenous without macro-segregation. Twenty point analyses were done along a line with 1 µm distance between the points, and gave the following RS alloy composition: for $Ti_{0.8}V_{0.2}$, Ti content was 80.0%±0.1% and the V content was 19.9% \pm 2%; while for Ti_{0.8}V_{0.1} the Ti content was 89.3% \pm 1% and the V content was 10.6% \pm 2%. This indicates that macroscopic inhomogeneity was not the case.

3.2 Hydrogenation properties

The hydrogenation kinetics for both $\mathrm{Ti}_{0.8}\mathrm{V}_{0.2}$ and $Ti_{0.9}V_{0.1}$ was excellent. The activated alloys were able to absorb hydrogen within 1 min with H sorption capacities depending on the hydrogenation temperature. At room temperature, the hydrogen capacity of both alloys was 3.95% H (mass fraction). This value slightly decreased to (3.0%-3.7%) H when the hydrogenation was carried out at 450 °C due to the thermodynamic limitations. Rapidly solidified alloys needed a short incubation time prior to hydrogen absorption, as seen in Fig. 4. In the case of the rapidly solidified Ti_{0.9}V_{0.1}, activated ribbons needed about 30 s before hydrogen started to interact with the alloy. The second rehydrogenation kinetics of the RS alloys was very fast and slightly faster than that of the as-cast alloys. Similar incubation time was needed for $Ti_{0.8}V_{0.2}$. The reason for the incubation time must be the formation of an oxide layer on the ribbon surface preventing hydrogen dissociation at the beginning of the hydrogen absorption process. However, fast kinetics can be achieved after rehydrogenation, when the ribbons have been fractured and new surfaces have been created facilitating hydrogen penetration into the bulk material. However, the difference in absorption kinetics between the as-cast and rapidly solidified alloys seems to be marginal as both cast and RS alloys are able to fully absorb hydrogen within less than 60 s.



Fig. 2 Microstructures of rapidly solidified alloys: (a) RS $Ti_{0.8}V_{0.2}$ (cross-section view); (b) High magnification image of RS $Ti_{0.8}V_{0.2}$; (c) RS $Ti_{0.9}V_{0.1}$ (cross section view); (d) High magnification view of RS $Ti_{0.9}V_{0.1}$ showing a plate-like submicron structure



Fig. 3 Microstructures of $Ti_{0.8}V_{0.2}$ and $Ti_{0.9}V_{0.1}$ alloys and their vanadium mapping: (a, b) RS $Ti_{0.8}V_{0.2}$; (c, d) RS $Ti_{0.9}V_{0.1}$ (The amount of vanadium is indicated by the colour scale on right side of the maps)



Fig. 4 Hydrogen absorption curves for RS $Ti_{0.9}V_{0.1}$ at 100 °C and 4 bar initial hydrogen pressure in comparison with as-cast alloy

TDS traces of hydrogen desorption from the fully hydrogenated samples are shown in Fig. 5. All traces have common features. They contain two well resolved desorption peaks. As expected, the increase of vanadium content reduces the thermal stability of the hydrides; indeed, the temperature of the main desorption peak shifts from around 500 °C for the hydride of as-cast $Ti_{0.9}V_{0.1}H_2$ to 460 °C for $Ti_{0.8}V_{0.2}H_2$ also synthesized from the as-cast alloy. Rapid solidification significantly destabilizes $Ti_{0.8}V_{0.2}H_2$ hydride, with the main desorption peak shifting by 50 °C, down to 410 °C. In contrast, RS



Fig. 5 TDS traces of hydrogen desorption from $Ti_{0.9}V_{0.1}H_2$ and $Ti_{0.8}V_{0.2}H_2$ dihydrides

does not affect the stability of the $Ti_{0.9}V_{0.1}H_2$ where the location of the peak does not change, being situated at 500 °C for both as-cast and RS hydrogenated samples. From the data given in Fig. 5, it can be concluded that kinetics of hydrogen desorption from the hydrides is improved for the rapidly solidified samples. Indeed, hydrogen desorption flows are always higher for the RS alloy hydrides.

The destabilization of the RS $Ti_{0.8}V_{0.2}H_2$ hydride observed in the TDS study can be linked to the changes in the thermodynamic and/or kinetic properties. From the thermodynamic point of view, the destabilization of the hydride may be due to a compositional change in the Ti_{0.8}V_{0.2} alloy during the RS process. Even though, EPMA does not show any evidence of macro-segregation, it is possible that the alloy contains small regions enriched in V, namely Ti_{0.8-x}V_{0.2+x} formed on a micro scale level; corresponding hydride has a lower stability as compared to the equilibrium Ti_{0.8}V_{0.2}H₂. This proposal is in agreement with the X-ray diffraction data (see the next section), showing that phase composition and the element distribution in the alloys are altered by the rapid solidification process. The second possible reason is the reduced crystallite size of the RS samples yielding improved kinetics of hydrogen exchange. We suggest that both mentioned factors contribute to the destabilization of the Ti_{0.8}V_{0.2} hydride. However, the situation is more complex for the RS Ti_{0.9}V_{0.1}. Here, despite lowering of the crystallite size on rapid solidification, a destabilization was not observed as compared to the as-cast sample.

3.3 Crystal structures

Phase-structural composition and crystallographic data derived from the Rietveld refinements of the SR-XRD pattern for the as-cast and RS alloys are listed in Table 1. The as-cast Ti_{0.8}V_{0.2} alloy after one complete hydrogen absorption–desorption cycle contained mainly the β -BCC solid solution Ti–V (97%, mass fraction) and α -Ti (3%). The RS Ti_{0.8}V_{0.2} alloy contained a lower amount of the BCC β -phase (90.3%); while the amount of the α -phase increased to 7%. Unit cell parameter of the β phase of the as-cast alloy, 3.2322(4) Å, is higher as compared to that of the RS Ti_{0.8}V_{0.2}, 3.216 Å. This data evidenced lowering of Ti content in the β phase during the rapid solidification, as the amount of α -Ti increased from 3% to 7% (mass fraction). Crystallite size derived

from the refinements of the profiles of XRD pattern gave $1.53 \mu m$ for the as-cast alloy, while much refined value of 210 nm was observed for the dehydrogenated rapidly solidified sample.

The ex situ SR-XRD powder diffraction pattern of the as-cast $Ti_{0.9}V_{0.1}$ alloy obtained after one complete hydrogen absorption-desorption cycle is shown in Fig. 6(a). The alloy contained a mixture of the β and α phases in a ratio of 32/67. During the rehydrogenation, the BCC phase formed a FCC dihydride, see SR-XRD pattern in Fig. 6(b). The RS Ti_{0.9}V_{0.1} contained 63% (mass fraction) β -phase which is almost twice more than that for the corresponding as-cast alloy. The unit cell parameter of the β -Ti_{0.9}V_{0.1} is lower than that of the β -phase in the Ti_{0.8}V_{0.2}. This must be due to the lower Ti content in the former phase. The unit cell parameter of the β -phase of the RS alloy is higher than that of the as-cast alloy due to the increased content of titanium in the RS alloy. The mean crystallite size derived from the refinements of the profile parameters equals 265 nm which is significantly lower as compared to the as-cast alloy, being 1.4 µm.

All studied alloys, as-cast and rapidly solidified, formed γ -FCC dihydrides, with space group $Fm\bar{3}m$, after interacting with hydrogen gas. Typical powder X-ray diffraction pattern of the FCC dihydride is shown in Fig. 6(b) presenting the data for the hydrogenated as-cast Ti_{0.9}V_{0.1}. The hydrogen content of this sample was close to 2 H/M (3.95% H) (mass fraction), and the unit cell parameter of the FCC dihydride was 4.4390(7) Å. Increased fraction of the β -phase as well as the decreased grain size observed in the RS Ti_{0.9}V_{0.1} shows a beneficial effects of the rapid solidification on the synthesis of the Ti_{0.9}V_{0.1} alloy for hydrogen storage applications.

Alloy	Phase (space group)	Abundance (mass fraction)/%	Unit cell parameter/Å	$V/\text{\AA}^3$
As-cast Ti _{0.8} V _{0.2}	$\beta - Ti_{0.8}V_{0.2}(Im\bar{3}m)$	98.55(5)	<i>a</i> =3.2322(4)	33.76(1)
	α -Ti(V) (<i>P</i> 6 ₃ / <i>mmc</i>)	1.4(1)	a=2.956(2) c=4.69(1)	35.55(8)
RS Ti _{0.8} V _{0.2}	β -Ti _{0.8-x} V _{0.2+x} ($Im\bar{3}m$)	90.8(8)	<i>a</i> =3.2165(2)	33.27(9)
	α -Ti(V) ($P6_3/mmc$)	7.7(5)	a=2.949(1) c=4.727(2)	35.62(5)
	VO _{2.5} (<i>I</i> 4/ <i>mmm</i>)	1.4(1)	a=2.976(1) c=3.477(4)	30.80(7)
As-cast Ti _{0.9} V _{0.1}	$\beta - Ti_{0.9}V_{0.1}(Im\bar{3}m)$	32.3(4)	<i>a</i> =3.20770(27)	33.005(5)
	α -Ti(V) (<i>P</i> 6 ₃ / <i>mmc</i>)	67.5(3)	a=2.94272(21) c=4.6798(4)	35.097(5)
RS Ti _{0.9} V _{0.1}	$\beta - Ti_{0.9}V_{0.1}(Im\bar{3}m)$	63.0(10)	<i>a</i> =3.272(1)	33.61(5)
	α -Ti(V) (<i>P</i> 6 ₃ / <i>mmc</i>)	32.7(8)	a=2.933(1) c=4.665(2)	34.77(3)
	$TiO_2(P4_2/mnm)$	3(1)	a=4.708(9) c=2.81(2)	62.3(5)

 Table 1 Phase-structural composition and crystallographic data for dehydrogenated alloys determined from Rietveld refinements of ex situ SR-XRD data



Fig. 6 Powder diffraction pattern and their Reitveld refinements: (a) Dehydrogenated as-cast $Ti_{0.9}V_{0.1}$; (b) Fully rehydrogenated sample forming FCC dihydrides

3.4 In situ SR-XRD study

Pd-modified dihydride of the hydrogen saturated as-cast Ti_{0.9}V_{0.1} alloy (Ti_{0.9}V_{0.1}H₂ + nano Pd) was used for the in situ SR-XRD studies of hydrogen desorption and absorption. During the diffraction experiments, the sample was heated and cooled with programmed rate of the temperature changes in controlled gas environment. In this experiment, the total time needed to perform one scan was about 95 s, so that the temperature change during that time was about 8 °C. Hydrogen desorption experiments were done by heating the sample from RT to 250 °C with heating rate of 20 °C/min and then, with a lower heating rate of 5 °C/min to reach a setpoint temperature of 800 °C; the experiments were performed in a flow of hydrogen gas ($p = \sim 1.12$ bar).

Figure 7 shows a 2D plot of the diffraction pattern from the sample during the course of experiment as well as the temperature history. During the heating, at 500 °C hydrogen started to desorb from the γ -hydride of Ti_{0.9}V_{0.1}H₂; the desorption yielded β -phase and a small amount of α -phase. Further heating resulted in enhanced transformation of the γ hydride to β hydride; the decomposition of γ hydride was completed at 700 °C. Even though the initial sample (at 25 °C) can be fitted well with a single phase FCC hydride with a unit cell parameter of 4.4390(7) Å, the formation of α -phase from the γ -hydride at about 500 °C indicated that the sample may contain another FCC hydride. This must be the FCC TiH₂ decomposing to α -HCP phase on dehydrogenation [20].



Fig. 7 2D X-ray diffraction intensity plot vs scan number from in situ experiments on $Ti_{0.9}V_{0.1}H_2$ (hydrogenated as-cast alloy) during dehydrogenation (heating) and rehydrogenation (cooling) (The dashed line gives a temperature profile)

Prior to rehydrogenation, the sample was cooled from 800 °C to 640 °C. During this period the formation of α -phase from β -phase was observed. At 640 °C, the sample was subjected to a mixture of H₂+CO gases (mixture of 90/10) with flow rate 20 mL/min and outlet pressure 1.12 bar. The sample readily absorbed hydrogen yielding a FCC dihydride. This transformation was completed at about 380 °C; no further transformations were observed upon continuing cooling. As listed in Table 2, at the end of rehydrogenation, the sample was composed of 51% (mass fraction) γ hydride, 47% α phase and small amount of β phase. It is evident that the β phase was able to absorb hydrogen and form dihydrides. The α phase can be considered to be not active in hydrogen uptake at the applied experimental conditions. The reason for that can be two folds: (a) bulk absorption of oxygen which stabilized the HCP α -phase making its hydrogenation impossible; (b) blocking by CO molecules of the metal surface thus creating barriers for hydrogen adsorption and dissociation, and preventing its diffusion into the bulk.

Sample	Phase (space group)	Abundance (mass fraction)/%	Unit cell parameter/Å	V/Å ³
Initial sample at RT	γ -Ti _{0.9} V _{0.1} H _{~2} (<i>Fm</i> 3 <i>m</i>)	100	4.4390(7)Å	87.47(4)
	γ -Ti _{0.9} V _{0.1} H _{~2} (<i>Fm</i> 3 <i>m</i>)	50.8(1)	<i>a</i> =4.4301(6)	86.94(3)
After rehydrogenation at 280 °C	$\beta\text{-Ti}_{0.9-x}V_{0.1+x}(Im\bar{3}m)$	~2.3	<i>a</i> =3.296	36.6(2)
	α -Ti(V/H/O) (<i>P</i> 6 ₃ / <i>mmc</i>)	46.8(1)	<i>a</i> =2.9814(7) <i>c</i> =4.487(2)	37.31(1)

Table 2 Phase-structural composition and crystallographic data for initial saturated hydride and rehydrogenated $Ti_{0.9}V_{0.1}H_2$ sample obtained from Rietveld refinements of powder X-ray diffraction data

4 Conclusions

The alloy preparation process determined the microstructure, crystal structure and the hydrogen storage properties of the binary $Ti_{0.8}V_{0.2}$ and $Ti_{0.9}V_{0.1}$ alloys. As-cast $Ti_{0.8}V_{0.2}$ alloy was primarily composed of the β phase with BCC structure, while Ti_{0.9}V_{0.1} contained a mixture of α and β phases with a dominant phase being the α HCP one (67%). Rapid solidification refined the grain size and changed the abundance of the phase constituent. For the Ti_{0.8}V_{0.2}, RS decreased the grain size from an average size of above 500 µm to around 12 µm and destabilized its corresponding hydride. For the Ti_{0.9}V_{0.1}, RS refined the microstructure as well and doubled the β phase fraction as compared to the as-cast Ti_{0.9}V_{0.1}. From Rietveld profile refinements, it was concluded that the crystallite size of the RS alloys significantly decreased from 1.53 µm for the as-cast $Ti_{0.8}V_{0.2}$ alloy to 210 nm for the RS $Ti_{0.8}V_{0.2}$ and from 1.4 μ m for the as-cast Ti_{0.9}V_{0.1} to 265 nm for the RS Ti_{0.9}V_{0.1}. In situ SR-XRD experiments showed that when hydrogenation was performed in a mixture of H_2 +10% CO, the β phase was more reactive towards hydrogen absorption and was transformed into the FCC dihydride. On the other hand, the α phase was inactive towards hydrogen in such a gaseous mixture of H₂+10%CO. Thus, microstructures which contain mainly β phase are preferable to be used for hydrogen separation applications from gas mixtures containing hydrogen and CO. From this prospective, rapid solidification improves properties of the alloys by increasing the amount of β phase. Furthermore, RS refined the grain size, which resulted in a faster desorption kinetics from the $Ti_{0.8-0.9}V_{0.2-0.1}H_2$.

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快速凝固富钛 Ti-V 合金的微观组织和储氢性能

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摘 要:研究目的在于优化 Ti_{0.8-0.9}V_{0.2-0.1} 二元合金的相结构成分、微观组织和储氢性能。该合金主要用于从含有 大量一氧化碳的高温气态混合物中吸收氢气。Ti_{0.8-0.9}V_{0.2-0.1} 合金中的 α -(HCP) 和 β -(BCC) 相在纯氢气中基于氢化 作用,形成单相 FCC 结构的氢化物,此过程与合金的化学成分无关。同步辐射 X 射线衍射的原位分析表明,在 含有氢气和 10%一氧化碳的混合气体中,只有 β 相转变成相应的氢化物。快速凝固 (RS) 处理细化了 Ti_{0.8}V_{0.2}和 Ti_{0.9}V_{0.1} 合金的晶粒组织,而且,快速凝固处理增加了 Ti_{0.9}V_{0.1} 合金中的 β 相,其所占比例是普通熔铸条件下的两 倍。扫描电子显微镜 (SEM) 分析表明,Ti_{0.9}V_{0.1} 合金含有片状组织,层片的厚度约为 300 nm。热脱附谱 (TDS) 显 示,微观组织的细化可以加快氢脱附的动力学过程。

关键词:储氢;Ti-V 合金;快速凝固;同步辐射 X 射线衍射;一氧化碳

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