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Determination of Phase Equilibria among δ -Fe, γ -Fe and Fe_2M Phases in Fe-Cr-M (M: Hf, Ta) Ternary Systems

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Abstract: Phase equilibria among δ -Fe, γ -Fe, and Fe_2M phases in the Fe-Cr-M (M: Hf, Ta) ternary systems were determined using bulk alloys heat-treated at high temperatures. The final goal of this study is to develop novel ferritic heat resistant steels strengthened by precipitation of Fe_2M phase on the eutectoid type reaction path: $\delta \rightarrow \gamma + \text{Fe}_2\text{M}$. The phases present in heat-treated samples were identified by microstructural characterization and X-ray diffraction pattern analysis. The chemical compositions of the phases were analyzed by energy dispersive spectroscopy. A pseudo-eutectoid trough ($\delta \rightarrow \gamma + \text{Fe}_2\text{M}$) exists at $\sim 1220^\circ\text{C}$ at a Hf content of 0.1% and at $\sim 1130^\circ\text{C}$ at a Ta content of 0.6% on the vertical section at a Cr content of 9.5% in each ternary system, respectively. A thermodynamic calculation with a database that reflects reported binary phase diagrams and the present study indicates that an increase in the Cr content decreases the temperature and the Hf/Ta contents of the pseudo-eutectoid troughs. The determined phase equilibria suggest that the supersaturation of Hf/Ta for the formation of γ phase is higher in the Hf doped system than in the Ta doped system, which is probably an origin of a much slower kinetics of precipitation on the eutectoid path in the latter system.

Citation: Yuan, Z.; Kobayashi, S. Determination of Phase Equilibria among δ -Fe, γ -Fe and Fe_2M Phases in Fe-Cr-M (M: Hf, Ta) Ternary Systems. *Metals* **2022**, *12*, 102. <https://doi.org/10.3390/met12010102>

Academic Editor:
Carlos Garcia-Mateo

Received: 10 December 2021

Accepted: 2 January 2022

Published: 5 January 2022

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Keywords: phase diagram; 9Cr-based heat-resistant ferritic steels; laves phase; interphase precipitation; fibrous precipitation

1. Introduction

Fossil fuel-fired steam and gas turbine power generations are currently supporting about 80% of the energy supply in the world [1]. Since the carbon dioxide emissions from the fuel combustion, however, a recent worldwide trend is to replace the fossil fuel-fired power generation with renewable energy such as solar power and wind power as the main power source in the near future [2]. Steam and gas turbine power generation systems, however, have advantages in providing energy security, including an ability to supply energy in case of an emergency and to back up the variable output from the renewable energy, and will thus be necessary also in future as important power sources while using clean fuels such as NH_3 and H_2 and/or carbon capture technologies. It is, therefore, important to continue make efforts to further increase the efficiency and the durability of steam/gas turbine power generation systems, thereby improving the high temperature durability of heat-resistant steels and alloys for the systems.

High Cr ferritic (martensitic) heat resistant steels are an important class of materials for high temperature components such as pipes and turbines in steam turbine power generation systems due to their low thermal expansion coefficient, high thermal conductivity, and relatively low cost compared with austenitic heat resistant steels and nickel based superalloys. The ferritic steels are designed in such a way that a ferritic matrix, which is originated from martensitic transformation, is strengthened with M_{23}C_6 type carbide (M: Cr, Fe, Mo) and (V, Nb)X carbonitride (X: C, N) precipitates against dislocation motion, recovery, and recrystallization. The strengthening precipitates are, however, prone to particle

coarsening and decomposition, which are assumed to be the main cause of the degradation of long-term creep strength of the steels [3,4].

Laves phase is an A_2B type intermetallic compound, which is known to form in many alloy systems including Fe based systems [5]. It is reported that Fe_2W and Fe_2Mo type Laves phases are precipitated during creep conditions within the ferritic matrix in conventional heat resistant ferritic steels [4]. The Laves phase precipitates are observed as fine particles on the lath/block boundaries and/or as coarse globular ones, depending on the type and the content of the Laves phase forming elements. Efforts have been made to improve the creep life of high Cr ferritic heat resistant steels by modification of Laves phase distribution and type. Positive effects on the creep resistance have been reported [6,7].

One of the present authors [8,9] found the formation of periodically arrayed rows of fine Fe_2Hf Laves phase particles in 9 wt. % Cr based ferritic matrix through interphase precipitation. This precipitation mode was identified to occur on a pseudo-eutectoid reaction path: $\delta\text{-Fe} \rightarrow \gamma\text{-Fe} + Fe_2Hf$, with a subsequent phase transformation of the γ phase into the $\alpha\text{-Fe}$ phase. Observed fine distribution of the Laves phase particles allows us to expect to develop a new type of heat resistant ferritic steel strengthened by the eutectoid type precipitation route. The precipitation kinetics of the Fe_2Hf type Laves phase and effects of the Laves phase on creep rupture strength have been investigated [10,11].

Fe_2Ta phase and Fe_2Nb phase is also available through the eutectoid reaction path: $\delta \rightarrow \gamma + Fe_2M$ (M : Ta, Nb) in the Fe-Ta and Fe-Nb binary systems, respectively [5]. Our preliminary experiments revealed that Fe_2Ta phase was formed with a fibrous morphology on the above type of reaction path in an Fe-Cr-Ta ternary alloy. The formation of fibrous Laves phase, hereafter called “fibrous precipitation” [12], was found to occur three orders of magnitude slower than interphase precipitation observed in Fe-Cr-Hf ternary alloys with similar Cr contents. Here a question arises regarding what criterion exists between the precipitation modes and if there is a correlation between the precipitation mode and the kinetics of the eutectoid reaction. An idea was proposed in steels [13] that precipitation mode in the eutectoid reaction path: $\gamma \rightarrow \alpha + \text{carbide}$ would be changed from fibrous precipitation to interphase precipitation by increasing the relative kinetics of $\gamma \rightarrow \alpha$ phase transformation (their interphase boundary migration) with respect to the growth rate of carbide particles. This speculation was, however, not confirmed by experiments and not evaluated on a quantitative basis. Reducing the Cr content in Fe-Cr-Hf/Ta alloys is expected to stabilize the γ phase with respect to the δ phase and thereby enhancing the rate of $\delta \rightarrow \gamma$ phase transformation against the formation of Laves phase, which might influence the precipitation mode on the eutectoid reaction. Knowledge of phase diagrams in the ternary systems are, therefore, needed to confirm the above hypothesis.

Addition of Hf to ferritic steels is reported to improve the corrosion resistance of the steels [14]. Ta addition was also known to increase the corrosion resistance and creep strength of ferritic heat resistant steels [15]. Those effects were found in steels with complex chemical compositions, but the knowledge of phase equilibria in the Fe-Cr-Hf/Ta ternary system would be the basis for understanding the microstructure and mechanical properties of the complex alloys.

Nevertheless, the knowledge of phase equilibria among the δ , γ , and Fe_2M phases in the two systems are limited. A vertical section of phase diagram in the Fe-Cr-Hf ternary system was reported but its dependence on the Cr content has not been clarified [8]. Calculated isothermal sections were reported in the Fe-Cr-Ta ternary system [16] but no experimental works have been reported as far as the authors' knowledge. The aim of this work is to determine the phase equilibria among the δ , γ , and Fe_2M phases in the Fe-Cr-Hf and Fe-Cr-Ta ternary systems experimentally. The results are used to calculate the vertical sections of phase diagram at different Cr contents to aim at understanding the basis for microstructural formation mechanisms along the pseudo-eutectoid reaction in the Fe based alloy systems for design of a new heat resistant ferritic steel.

2. Materials and Methods

The nominal compositions of the alloys investigated in this study are listed in Table 1 (hereafter all compositions are given in atomic percent unless specified otherwise). The choice of the chemical compositions is to have a mixture of δ/γ /Laves phases at a high temperature and thereby obtaining the equilibrium compositions of the three-phase tie-triangle in each system. These alloys were prepared by arc melting under an argon atmosphere. The starting materials used are as follows: electrolytic iron of 4N purity, chromium of 3N purity, hafnium of 3N purity, and tantalum of 3N purity. The alloys are designated with their Hf or Ta content. The ingots were cut to rectangular samples of 15 mm $\times$ 9 mm $\times$ 3 mm in size and heat-treated at various temperatures, followed by water quenching. The heat treatments were conducted in silica tubes vacuumed and backfilled with argon. The conditions of the heat treatments are shown in Table 1. Choice of the two different heat treatment times was intended to identify if the phase states are kept or changing with time, i.e., whether the alloys are in equilibrium or not.

Table 1. The nominal compositions of the alloys studied (compositions are given in atomic percent).

Designation	Nominal Composition/at. %				Heat Treatment Conditions (Temp./Time)
	Fe	Cr	Hf	Ta	
2 Hf	Bal.	9.8	2.0	-	1150 °C/48 h
4 Hf	Bal.	9.6	4.0	-	1150 °C/20 h, 48 h
2.5 Ta	Bal.	9.2	-	2.5	1100 °C/48 h, 235 h
5 Ta	Bal.	9.6	-	5.0	1100 °C/48 h, 235 h

The heat-treated samples were cut at cross sections 1 mm inside from the sample surfaces, and the cross sections were ion milled with Ar gas. The microstructures were examined by field emission scanning electron microscopy (FESEM, JSM-7000F, JEOL Ltd., Tokyo, Japan), which is equipped with a backscattered electron (BSE) detector. The phases present were identified by microstructural features and X-ray diffraction (XRD) analysis. The XRD was obtained with Cu radiation using a Ni filter (Mini Flex 600, Rigaku Corp., Tokyo, Japan). The chemical compositions of the phases were determined by energy dispersive spectroscopy (EDS) on FESEM. For the analysis, calibration curves were made to correlate the intensities of Fe, Cr, Hf, and Ta with their chemical compositions by using several heat-treated samples as standards with the assumption that the nominal compositions and the alloy compositions are equal. More than 20 measurements were performed for each phase and the average values were determined. Isothermal sections and vertical sections of phase diagrams in the systems were calculated using a commercial software Pandat (Version 2018, CompuTherm LLC, Middleton, WI, USA). In the calculation, ternary interaction parameters were modified while binary parameters were kept fit calculated tie-triangles to the experimentally obtained one in each ternary system.

3. Results and Discussions

3.1. Microstructures of Heat-Treated/Quenched Samples and Phase Identification

Figure 1 shows BSE images of the alloys heat-treated for longer time. In the 4 Hf alloy heat-treated at 1150 °C for 48 h, three types of micro-constituents are observed (Figure 1a). The first constituent is of a relatively dark region without appreciable orientation contrast in it. The second one exhibits a similar brightness to the first one but with fine orientation contrast within it. The fine orientation contrast imaged in the second constituent indicates that a phase transformation took place during quenching after the heat treatment. The third constituent is of bright globular particles. In the 5 Ta alloy heat-treated at 1100 °C for 235 h (Figure 1b), the three micro-constituents, as observed in the 4 Hf heat-treated alloy, are found while the fraction of the first constituent is low.

Figure 1. BSE images of the alloys studied: (a) 4 Hf alloy heat-treated at 1150 °C for 48 h, (b) 5 Ta alloy heat-treated at 1100 °C for 235 h. Vertical bright lines are originated from Ar milling. The regions denoted by γ was transformed to α -Fe during quenching from heat treatment temperatures.

Figure 2 shows XRD profiles obtained from the 4 Hf and 5 Ta alloy samples of which micrographs are presented in Figure 1. In the 4 Hf sample, the observed diffracted peaks are identified as those from $\alpha(\delta)$ -Fe and C14-Fe₂Hf. In the 5 Ta sample the diffracted peaks are identified as $\alpha(\delta)$ -Fe and C14-Fe₂Ta. The lattice parameters of the α/δ -Fe and C14 phases were estimated using Cohen's method [17], and the results are listed in Table 2. The identified phases by XRD combined with the observed microstructural features indicates that the first constituent is δ -Fe which were frozen from the heat-treated temperature, the second one is α -Fe grains or martensite which were transformed from γ -Fe during quenching from the heat-treated temperature, and the bright globular particles are of the C14 phases. It can therefore be stated that the 4 Hf and 5 Ta alloy exhibits three phases, δ -Fe, γ -Fe and C14, at 1150 °C and 1100 °C, respectively. The phases presented at the heat treatment temperatures for all the samples were examined and the results are listed in Tables 3 and 4.

Figure 2. XRD profiles obtained from the 4 Hf and 5 Ta alloys: (a) 4 Hf alloy heat-treated at 1150 °C for 48 h, (b) 5 Ta alloy heat-treated at 1100 °C for 235 h.

Table 2. The estimated lattice parameters of the phases present in the alloys after heat treatments.

Designation	Heat Treatment	Phase Present	Lattice Parameter (Å)	
			a	c
4 Hf	1150 °C/48 h	α/δ -Fe	2.873 (5)	-
		C14-Fe ₂ Hf	4.921 (0)	8.017 (7)
5 Ta	1100 °C/235 h	α/δ -Fe	2.874 (8)	-
		C14-Fe ₂ Ta	4.808 (1)	7.852 (3)

3.2. Chemical Analysis and Isothermal Sections

The chemical compositions of the δ -Fe, γ -Fe and C14-Fe₂M phases at heat treatment temperatures were determined and the results are listed in Tables 3 and 4. The solubility of Hf in the δ -Fe phase is ~0.1% and is lower than a detection limit (0.1%) in the γ -Fe phase (Table 3).

Table 3. The analyzed chemical compositions of the phases present in the Hf-doped ternary alloys after heat treatments, followed by water quenching.

Designation	Heat Treatment	Phase Present	Chemical Composition (at.%)		
			Fe	Cr	Hf
2 Hf	1150 °C/48 h	δ	Bal.	11.1	0.1
		γ	Bal.	9.9	~*
		Fe ₂ Hf	Bal.	4.3	24.9
4 Hf	1150 °C/20 h	δ	Bal.	11.1	0.1
		γ	Bal.	9.8	~*
		Fe ₂ Hf	Bal.	4.0	24.5
	1150 °C/48 h	δ	Bal.	11.1	0.1
		γ	Bal.	9.8	~*
		Fe ₂ Hf	Bal.	4.0	26.6

* Below the detection limit in EDS.

Table 4. The analyzed chemical compositions of the phases present in the Ta-doped ternary alloys after heat treatments, followed by water quenching.

Designation	Heat Treatment	Phase Present	Chemical Composition (at.%)		
			Fe	Cr	Ta
2.5 Ta	1100 °C/48 h	γ	Bal.	9.4	0.3
		Fe ₂ Ta	Bal.	5.5	27.0
	1100 °C/235 h	γ	Bal.	9.3	0.3
		Fe ₂ Ta	Bal.	5.6	26.5
5 Ta	1100 °C/48 h	δ	Bal.	10.1	0.6
		Fe ₂ Ta	Bal.	5.6	28.2
	1100 °C/235 h	δ	Bal.	10.2	0.5
		γ	Bal.	9.3	0.3
		Fe ₂ Ta	Bal.	5.6	27.7
		Fe ₂ Ta	Bal.	5.6	27.7

The solubility of Ta is higher in the δ -Fe phase than in the γ -Fe phase, as in the Hf case, but those of Ta in both phases are higher than those of Hf. Cr is enriched in the δ phase compared to the γ phase, as in the Fe-Cr binary alloy system. It is noted that the chemical compositions of the phases are almost the same in a heat treatment period between 20 h and 48 h in the 4 Hf alloy and between 48 h and 235 h in the 2.5 Ta alloy. This finding leads to a conclusion that a heat treatment time of 48 h in the former case and of 235 h in the latter case is sufficient to reach the equilibrium state at each heat treatment temperature, respectively. Absence of γ phase in the 5 Ta alloy heat treated for 48 h is possibly due

to a small equilibrium fraction of the phase and a slight compositional deviation in the sample towards the $\delta + \text{Fe}_2\text{Ta}$ two-phase region. The lower Hf or Ta contents in the Fe_2M phase observed in shorter heat-treated samples and in samples with lower Hf/Ta contents would be probably caused by their particle sizes smaller than the interaction volume in EDS measurements.

According to the composition analysis, the tie-triangle is plotted on the isothermal section of phase diagram at each heat treatment temperature in each alloy system and is shown in Figures 3 and 4. The compositions of the phases on the Fe-Cr, Fe-Hf, and the Fe-Ta binary systems are referred to the data in the literature [5]. No data were reported on the isothermal section of phase diagram in the Fe-Cr-Hf ternary system, and the result obtained in the present study is probably the first data. The solubility of Hf in the γ phase is found to decrease with increasing the Cr content by $\sim 10\%$, as can be seen on the Fe-rich portion (Figure 4a). The phase boundary between $\gamma + \text{Fe}_2\text{Hf}/\text{Fe}_2\text{Hf}$ phase regions is extended to an almost equi Cr content direction in the Fe-Cr-Hf ternary system (Figure 3a). Harikumar et al. calculated phase equilibrium in the Fe-Cr-Ta ternary system and reported the isothermal sections of phase diagram at several temperatures including 1150 °C and 1050 °C. The tie triangle of δ , γ and Fe_2Ta phases determined at 1100 °C in the present study is located at somewhat a lower Cr content than expected from the report by Harikumar et al. The locations of the tie triangles are compared in Figure 4b. This discrepancy cannot be explained by possible inclusion of a small amount of C and N in the present experimental alloys since those elements are γ -phase stabilizer. A qualitative explanation would be an existence of ternary interaction to stabilize the δ phase with respect to the γ phase. The phase boundary between the $\gamma + \text{Fe}_2\text{Ta}/\text{Fe}_2\text{Ta}$ phase regions is extended to the equi-Cr content direction at 1100 °C also in the Fe-Cr-Ta ternary system.

Figure 3. Isothermal sections of phase diagram showing the tie-triangle of δ , γ and Fe_2M in the Fe-Cr-M (M: Hf, Ta) ternary systems: (a) Fe-Cr-Hf, 1150 °C, (b) Fe-Cr-Ta, 1100 °C.

3.3. Calculation of Vertical Sections of Phase Diagram and Confirmation by Experiments

Vertical sections of phase diagram at different Cr contents are calculated using a thermodynamic database in which ternary interaction parameters were modified to fit the calculated tie-triangle of δ , γ and Fe_2M to the experimentally obtained one. Figure 5 shows calculated vertical sections at a constant Cr content of 9.5% and of 7.5% in the two ternary systems. The pseudo-eutectoid reaction trough can be found at all the sections. The trough is located at $\sim 0.1\%$ Hf and 1220 °C on the 9.5% Cr section. It moves towards the higher Hf content and higher temperature as the Cr content is decreased.

Figure 4. Isothermal sections of phase diagram in Fe-rich portion in the Fe-Cr-M (M: Hf, Ta) ternary systems: (a) Fe-Cr-Hf, 1150 °C, (b) Fe-Cr-Ta, 1100 °C. Tie-triangles at 1150 °C and 1050 °C reported in [14] are included in (b).

Figure 5. Calculated vertical sections of phase diagram in the Fe-Cr-M (M: Hf, Ta) ternary systems: (a) Fe-9.5Cr-Hf, (b) Fe-7.5Cr-Hf, (c) Fe-9.5Cr-Ta, (d) Fe-7.5Cr-Ta. Eutectoid coordinates are designated in each section. Experimentally determined phase fields are also shown.

In the Fe-Cr-Ta ternary system, the pseudo eutectoid trough is located at ~0.6% Ta and 1130 °C and moves towards the higher Ta content and higher temperature with decreasing the Cr content, as in the Fe-Cr-Hf system. When compared at the same Cr contents, the Hf/Ta content and the temperature of pseudo-eutectoid trough is higher and lower, respectively, in the Fe-Cr-Ta system than the Fe-Cr-Hf system.

The validity of the calculated vertical sections was confirmed by heat treatments and microstructural observations on alloys with on the sections at the Cr content of 9.5% Cr. The results and phase boundary information in binary systems are shown in Figure 5a,c. A quite good agreement between the calculation and the experiments is recognized.

A supersaturation value, Ω , of Hf/Ta for the formation of γ and Fe_2M phases along the pseudo-eutectoid reaction path: $\delta \rightarrow \gamma + \text{Fe}_2\text{M}$ is calculated using the following Equations (1) and (2), respectively:

$$\Omega_{\delta \rightarrow \gamma} = (x_0 - x_\gamma) / (x_\delta - x_\gamma) \quad (1)$$

$$\Omega_{\delta \rightarrow \text{Fe}_2\text{M}} = (x_0 - x_\delta) / (x_{\text{Fe}_2\text{M}} - x_\delta) \quad (2)$$

where x_0 is the initial Hf/Ta content, x_γ , x_δ , and $x_{\text{Fe}_2\text{M}}$ the Hf/Ta content in each phase which can be calculated by assuming a two-phase equilibrium between δ and γ for (1) and between δ and Fe_2M for (2). The estimated Ω values for 0.1% Hf and 0.6% Ta at different Cr contents are shown in Figure 6. The dependence of $\Omega_{\delta \rightarrow \gamma}$ values on the Cr content is higher in the Fe-Cr-Ta system than the Fe-Cr-Hf system, which results in much lower values in the former system at 9.5% Cr. The $\Omega_{\delta \rightarrow \text{Fe}_2\text{M}}$ values slightly increase with decreasing the temperatures and more than one order of magnitude smaller than the $\Omega_{\delta \rightarrow \gamma}$ values.

Figure 6. Calculated supersaturations Ω of M element on the pseudo-eutectoid reaction path: $\delta \rightarrow \gamma + \text{Fe}_2\text{M}$ (M: Hf, Ta): (a) Ω for the formation of γ phase in the Fe-xCr-0.1 Hf alloys ($x = 9.5, 7.5$), (b) Ω for the formation of γ phase in the Fe-xCr-0.6 Ta alloys ($x = 9.5, 7.5$), (c) Ω for the formation of Fe_2M phase in Fe-xCr-0.1 Hf alloys and Fe-xCr-0.6 Ta alloys ($x = 9.5, 7.5$).

3.4. Formation of Fe_2M Phase along the Pseudo-Eutectoid Reaction Path

As explained in introduction, Fe_2Ta type Laves phase is formed with a different morphology and much slower kinetics in Fe-Cr-Ta alloys than Fe_2Hf type Laves phase in Fe-Cr-Hf ternary alloys with a similar Cr content. Figure 7 compares precipitation modes of the Laves phases along the pseudo-eutectoid reaction path in a Hf doped and a Ta doped alloy with similar Cr contents. In the Fe-9.4Cr-0.1Hf alloy (Figure 7a), periodically arrayed rows of fine particles (Fe_2Hf) are formed on rapid cooling. The arrangement of the fine particles along the direction parallel to an interphase boundary, on which ledges are found, allows us to assume that the particles were formed by precipitation on interphase boundaries which move by a ledge migration, which is called interphase precipitation. It was found in our previous work that the interphase precipitation takes place at a short time period from a few seconds to a few tens of seconds [9,10]. In the Fe-9.7Cr-0.6Ta alloy (Figure 7b), fibrous precipitates (Fe_2Ta) are arranged perpendicular to an advancing interphase boundary. This precipitation mode, called fibrous precipitation, was found to occur in a time period around a few thousand seconds, which is three orders of magnitude slower than the interphase precipitation observed in the Hf-doped ternary alloys with a similar Cr content. The much slower kinetics in the Ta-doped alloy would be originated by lower $\Omega_{\delta \rightarrow \gamma}$ values in the alloy (Compare Figure 6a,b). This assumption and reason for the difference in precipitation mode between the alloys will be answered by further studies on alloys with controlled $\Omega_{\delta \rightarrow \gamma}$ values with changed Cr contents.

Figure 7. BSE images showing different types of precipitation modes for Fe_2M phases on the pseudo-eutectoid reaction path: $\delta \rightarrow \gamma + \text{Fe}_2\text{M}$: (a) interphase precipitation observed in the Fe-9.4Cr-0.1Hf alloy water-quenched from 1300 °C, (b) fibrous precipitation observed in the Fe-9.7Cr-0.6Ta alloy heat treated at a rate of 0.5 °C/s from 1250 °C. The γ phase in the transformed regions is transformed to α -Fe phase during cooling from high temperatures.

4. Conclusions

Phase equilibria among δ -Fe, γ -Fe, and Fe_2M phases in the Fe-Cr-M (M: Hf, Ta) ternary systems were determined using bulk alloys heat-treated at high temperatures in the context of developing novel ferritic heat resistant steels strengthened by precipitation of Fe_2M phase on a eutectoid type reaction path: $\delta \rightarrow \gamma + \text{Fe}_2\text{M}$. The main results are:

1. A pseudo-eutectoid trough ($\delta \rightarrow \gamma + \text{Fe}_2\text{M}$) exists at ~1220 °C at a Hf content of 0.1% and at ~1130 °C at a Ta content of 0.6% on the vertical sections at a Cr content of 9.5% in each ternary system, respectively;
2. Thermodynamic calculation with a database based on reported binary phase diagrams and the present study indicates that reducing the Cr content in the ternary alloy systems increases the temperature and the Hf/Ta contents of the pseudo-eutectoid troughs;
3. The determined phase equilibria suggest that the supersaturation of Hf/Ta for the formation of γ phase is higher in the Hf doped system than in the Ta doped system

at a Cr content of 9.5%, which is probably an origin of a much slower kinetics of precipitation on the eutectoid path in the latter alloy system.

Author Contributions: Conceptualization, S.K.; investigation, Z.Y.; writing—original draft preparation, Z.Y. and S.K.; writing—review and editing, Z.Y. and S.K. All authors have read and agreed to the published version of the manuscript.

Funding: This research was funded by Grant-in Aid for Scientific Research (C) in Japan Society for the Promotion of Science, grant number [15K06496], and a research funding by JFE 21st Century Foundation.

Data Availability Statement: Not applicable.

Acknowledgments: The authors thank Chong Jin of Tokyo Institute of Technology for his experimental assistance.

Conflicts of Interest: The authors declare no conflict of interest.

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