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Al-Fe-Ni (Aluminum-Iron-Nickel)

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In view of its technological importance, this system has been investigated experimentally a number of times. Many reviews and updates have also appeared in the last two decades [1988Ray, 1992Bud, 1994Rag, 1995Vil, 2005Cac, 2005Rag, 2006Ele, 2006Rag, 2008Rag, 2009Rag]. For a recent detailed summary, the reader is referred to the updated version of [1992Bud] by [2005Cac], the review of [2006Ele] and the updates by [2008Rag] and [2009Rag]. Recently, [2007Zha] carried out a thermodynamic analysis of the system covering the entire composition and a wide temperature range. This work was left out inadvertently in the update by [2009Rag]. It is reviewed here briefly.

Binary Systems

The Al-Fe phase diagram [1993Kat] shows that the face-centered cubic (fcc) solid solution based on Fe is restricted by a γ loop. The body-centered cubic (bcc) solid solution exists in the disordered *A*2 form (α), as well as the ordered *B*2 and $D0_3$ forms. Apart from the high temperature phase ε ($D8_2$, Cu_5Zn_8-type cubic; stable between 1232 and 1102 °C), there are three other intermediate phases in this system: $FeAl_2$ (triclinic), Fe_2Al_5(70-73 at.% Al, orthorhombic), and $FeAl_3$ or Fe_4Al_{13} (74.5-76.6 at.% Al, monoclinic). The Al-Ni phase diagram [1993Oka] shows five intermediate phases: $NiAl_3$ ($D0_{11}$, Fe_3C-type orthorhombic), Ni_2Al_3 ($D5_{13}$-type hexagonal), NiAl (*B*2, CsCl-type cubic, denoted β), Ni_5Al_3 (Ga_3Pt_5-type orthorhombic) and Ni_3Al ($L1_2$, $AuCu_3$-type cubic, denoted γ′). The Fe-Ni phase diagram [1993Swa] is characterized by a very narrow solidification range with a peritectic reaction at 1514 °C, between bcc δ and liquid that yields the Fe-based fcc solid solution. A continuous fcc solid solution denoted γ is stable over a wide range of temperature. At 517 °C, an ordered phase $FeNi_3$ forms congruently from γ.

Thermodynamic Assessment of [2007Zha]

[2007Zha] reviewed briefly the experimental data from the literature both on phase equilibria and thermodynamic properties. Selected data from the phase equilibria and thermodynamic measurements were used in the optimization. Recent results from [2008Zha] (reviewed by [2009Rag]) were included. In addition, new results from differential thermal analysis carried out by [2007Zha] on the alloys used by [2008Zha] were included. The data of [2007Chu1], [2007Chu2] and [2008Chu], reviewed by [2008Rag] and [2009Rag], were not used in the optimization.

[2007Zha] accepted the binary descriptions of the boundary systems in the literature with some slight modifications. The Gibbs energy of the ternary liquid phase was described by a Redlich-Kister polynomial. The binary compounds showing both a homogeneity range in the binary systems as well as a ternary extension were described by appropriate sublattice models. A single Gibbs energy expression that incorporates an ordering term was used for describing the ordered and disordered states of bcc (*A*2 and *B*2) and of fcc (*A*1 and $L1_2$). Among the ternary compounds, the decagonal quasicrystalline phase D (denoted τ_3 by [2007Zha]) was taken to be stoichiometric at $Al_{71}Fe_5Ni_{24}$. The τ_1 phase (denoted τ_2 by [2007Zha]) was modeled at constant Al content as $Al_5(Fe,Ni)_2$. The τ_2 phase (denoted τ_1 by [2007Zha]) was also modeled at constant Al content as $Al_9(Fe,Ni)_2$. The interaction parameters obtained from optimization were listed.

The isothermal sections at 1250, 1050, 950 and 850 °C computed by [2007Zha] are shown in Fig. 1-4 respectively. The ternary phase τ_1 ($Al_{10}Fe_3Ni$) has a stability range of 1182-627 °C and is present in Fig. 2-4. The ternary phase τ_2 (Al_9FeNi) is stable only below 805 °C and is not present in Fig. 1-4. The decagonal phase D forms at 926 °C and decomposes at 850 °C. At 1250 °C (Fig. 1), the bcc → B2 transition is second order. At 1050 °C (Fig. 2) and 800 °C (Fig. 3), this ordering transition is partly second-order and partly first-order.

Comparison with the experimental data from different sources showed good agreement with the calculated equilibria. For better clarity, the experimental data points are not

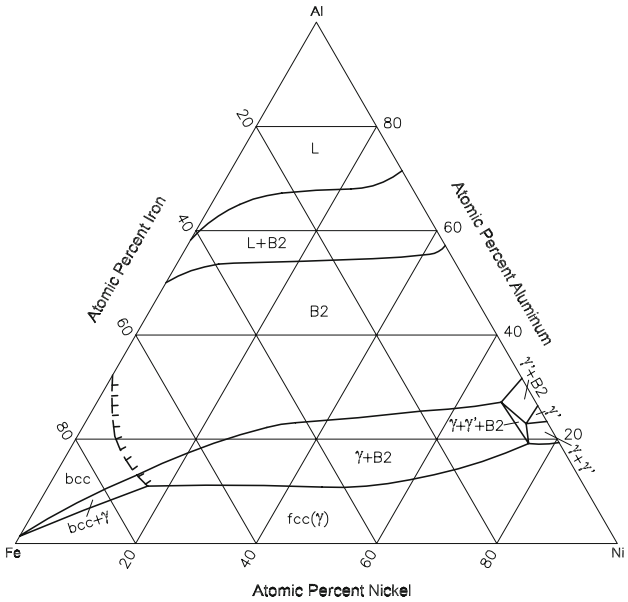

Fig. 1 Al-Fe-Ni computed isothermal section at 1250 °C [2007Zha]

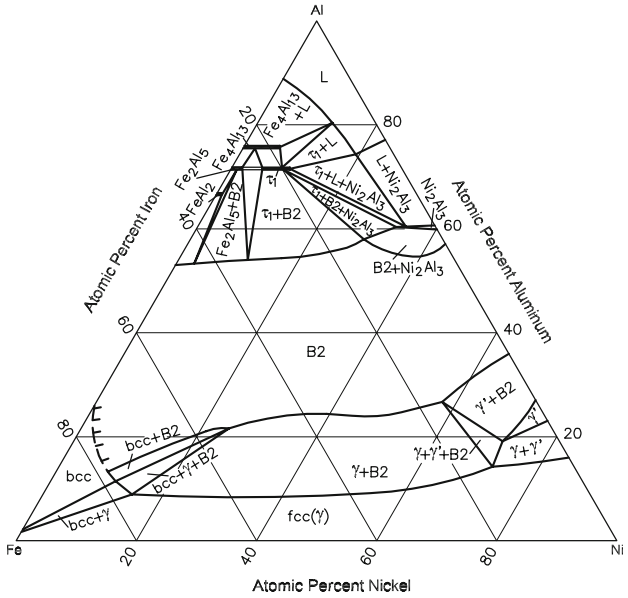

Fig. 2 Al-Fe-Ni computed isothermal section at 1050 °C [2007Zha]

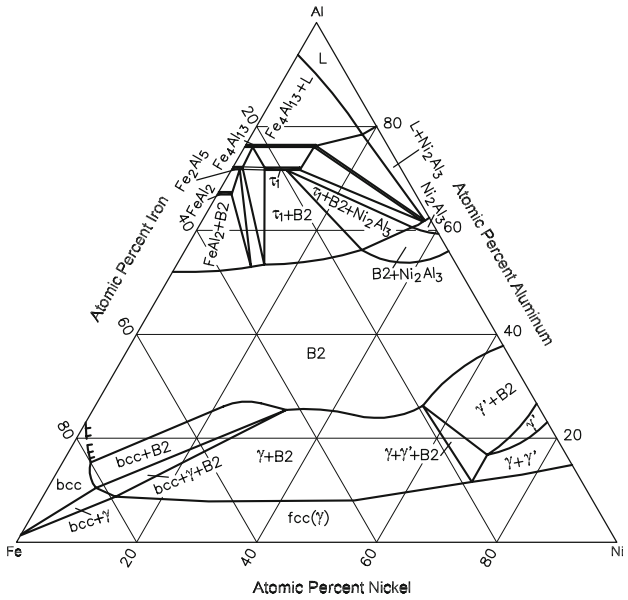

Fig. 3 Al-Fe-Ni computed isothermal section at 950 °C [2007Zha]

shown in Fig. 1-4. An exception to the general agreement was found in the extent of the solubility of Fe and Ni in the Al-rich liquid at 1050 °C. The calculated solubility was significantly lower than that found experimentally.

Three vertical sections at 1 mass% of Fe, 1 mass% Ni and along the Ni_3Fe-Ni_3Al join respectively were also computed and compared with experimental data included (as well as excluded) in the optimization. The agreement was found to be satisfactory. The pseudobinary region of the vertical section along the Ni_3Fe-Ni_3Al join computed by

Fig. 4 Al-Fe-Ni computed isothermal section at 850 °C [2007Zha]

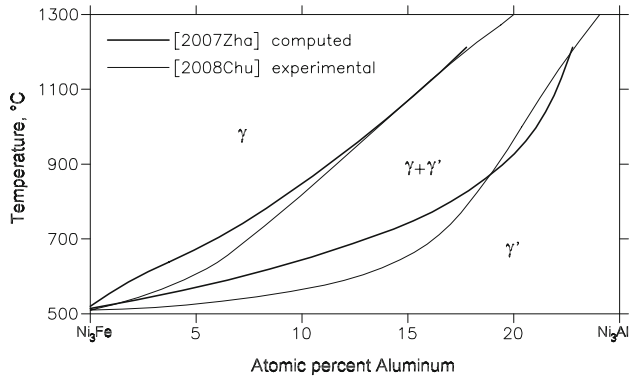

Fig. 5 Al-Fe-Ni partial vertical section along the Ni_3Fe-Ni_3Al join

[2007Zha] is compared with the experimental section of [2008Chu] in Fig. 5.

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