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Al-Mo-Ni (Aluminum-Molybdenum-Nickel)

V. Raghavan

The previous review of this system by [1993Kub] presented a reaction sequence, a liquidus surface, and isothermal sections at 1200, 1100, 930, 880, and 700 °C for Al-poor alloys from [1988Mas], a calculated full section at 1200 °C from [1974Kau], and two vertical sections at 14 at.% Al and 65 at.%, Ni respectively. The compilation by [1995Vil] gave a partial liquidus projection from [1983Nas], computed full isothermal sections at 1727, 1527, and 1200 °C from [1974Kau], partial isothermal sections for Al-poor alloys at 1390, 1380, 1360, 1340, and 1300 °C from [1986Mas], at 1260, 1171, 1093, 1038, and 947 °C from [1984Mir], and at 1200, 1100, 880, and 700 °C from [1988Mas], and full experimental isothermal sections at 950 °C from [1969Vir] and at 800 °C from [1969Mar]. Recent work reviewed here includes a full isothermal section from [1999Lu] at 1100 °C and partial sections for Al-rich alloys at 1050, 1000, 950, and 800 °C from [2002Gru].

Binary Systems

The Al-Mo phase diagram [1997Sau], with an added update for the Al-Mo_3Al_8 region by editor Smith, shows the following intermediate phases: $MoAl_{12}$ (WAl_{12}-type cubic), $MoAl_5$ (with three modifications), Mo_5Al_{22}, Mo_4Al_{17}, $MoAl_4$ (monoclinic), $Mo_{1-x}Al_{3+x}$ (cubic), $MoAl_3$ (monoclinic), Mo_3Al_8 (monoclinic), MoAl [body-centered cubic (bcc)], and Mo_3Al (*A*15, Cr_3Si-type cubic). The Al-Ni phase diagram [1993Oka] shows five intermediate phases: $NiAl_3$ (Fe_3C-type orthorhombic), Ni_2Al_3 ($D5_{13}$-type hexagonal), NiAl (*B*2, CsCl-type cubic, also denoted β), Ni_5Al_3 (Ga_3Pt_5-type orthorhombic), and Ni_3Al ($L1_2$, $AuCu_3$-type cubic; also denoted γ′). The Mo-Ni phase diagram [1991Oka] has three intermediate phases: Ni_4Mo ($D1_a$-type tetragonal), Ni_3Mo ($D0_a$-type orthorhombic), and NiMo (orthorhombic).

Ternary Phases

No ternary compounds are known in the Ni-rich region of this system. In the Al-rich region, two ternary phases are known. The N phase $Mo(Al_{1-x}Ni_x)_3$ (denoted τ_1 by [1993Kub]) is $TiAl_3$-type tetragonal, with $a = 0.3761$ nm and $c = 0.8412$ nm [2002Gru]. It has a Ni content of 1.6 to 6.0 at.% in as-cast condition and a smaller Ni range between these limits in the equilibrated condition. It is stable up to 1500 °C and down to <950 °C [2002Gru]. The X phase $Al_{75.1}Ni_{14.2}Mo_{10.7}$ (denoted τ_2 by [1993Kub]) had orthorhombic symmetry with a possible space group of *Pbm*2, *Pb*2$_1$*m*, or *Pbmm*. The lattice parameters are: $a = 1.0054$ nm, $b = 1.5288$ nm, and $c = 0.8519$ nm. It melts incongruently at 1015 °C [2002Gru].

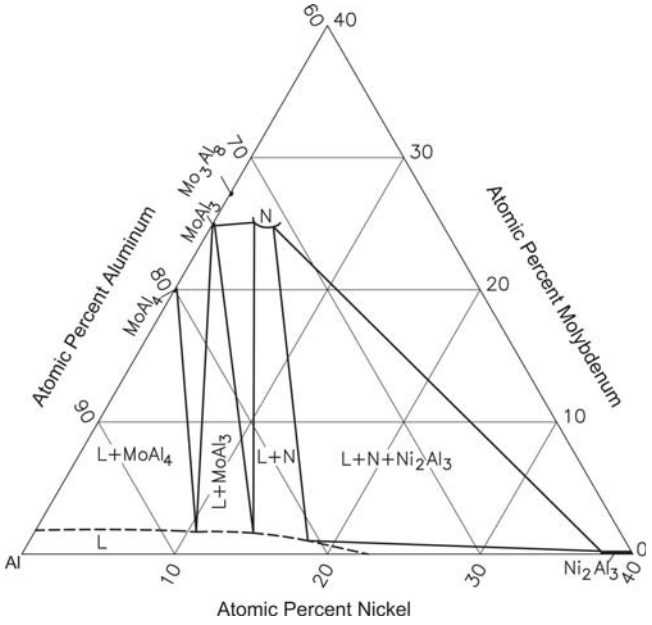

Fig. 1 Al-Mo-Ni partial isothermal section at 1050 °C [2002Gru]

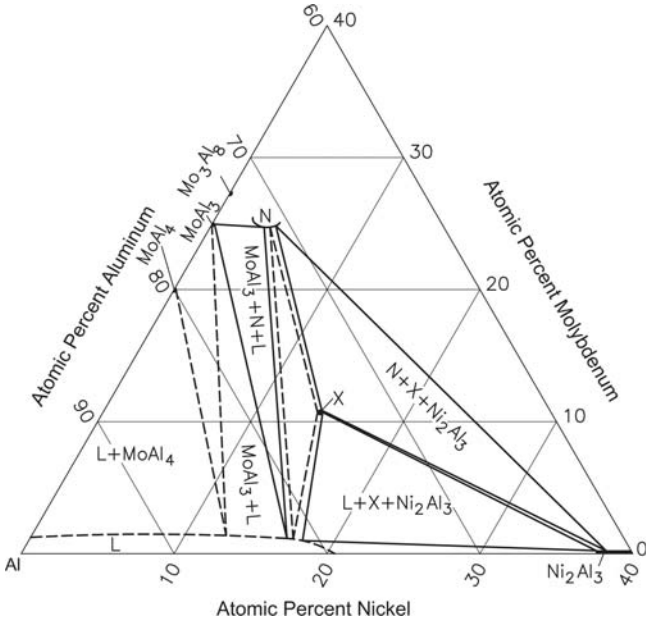

Fig. 2 Al-Mo-Ni partial isothermal section at 1000 °C [2002Gru]

Ternary Isothermal Sections

With starting metals of Al (99.9999 wt.%), Mo (99.98 wt.%), and Ni (99.98 wt.%), [2002Gru] prepared 12 ternary alloys by levitation melting. Annealing was done for up to 100 h in the range 1050 to 800 °C. The phase equilibria were studied by scanning and transmission electron microscopy, x-ray powder diffraction, and differential thermal analysis. Compositions of the phases were measured by the energy dispersive x-ray analysis. Partial isothermal sections were constructed by [2002Gru] at 1050, 1000, 950, and 800 °C. At 1050 °C (Fig. 1), the X phase is not stable. The solubility of Mo in Ni_2Al_3 is less than 0.1 at.% and of Ni in $MoAl_3$ and $MoAl_4$ is less than 0.3 at.%. Although the N phase, with the nominal formula $Mo(Al,Ni)_3$, is close to the Al-Mo side, no indication was found that it can be a ternary extension of a binary Al-Mo phase. The nearby Al-Mo phases $MoAl_3$ and Mo_3Al_8 have essentially different diffraction patterns and were distinguished by [2002Gru]. At 1000 °C (Fig. 2), the X-phase is present in addition to the N phase. It forms tie-lines with Ni_2Al_3, liquid, and the N phase. Between 1000 and 950 °C, two successive U-type transition reactions occur: $N + L \leftrightarrow MoAl_3 + X$ and $L + MoAl_3 \leftrightarrow MoAl_4 + X$. This results in the phase distribution at 950 °C seen in Fig. 3. At 800 °C (Fig. 4), equilibrium could not be achieved in the N phase region after annealing up to 90 h and the phase relationships in this region were not given by [2002Gru]. The three-phase equilibria of $X + L + Ni_2Al_3$ and $X + L + MoAl_5$ were found at 860 °C by [2002Gru]. This implies that the peritectic nucleation of $NiAl_3$ along the Al-Ni side at 854 °C is followed by the transition reaction: $L + Ni_2Al_3 \leftrightarrow X + NiAl_3$ to yield the phase distribution seen in Fig. 4.

With starting metals of Al (99.99 wt.%), Ni (99.97 wt.%), and Mo (99.97 wt.%), [1999Lu] arc melted two binary alloys with 51at.%Al-49at.%Ni and 73.4at.%Al-26.6at.%Mo, respectively. Diffusion triples were formed with each of the binary alloys, pure Ni, and pure Mo, and heat treated at 1100 °C. The compositions of the coexisting phases near the interfaces were measured by electron probe microanalysis. Tie-line data for the two-phase equilibria of (Ni)-Ni_3Al, Ni_3Al-NiAl, Ni_3Al-NiMo, and (Ni)-NiMo and the three-phase equilibrium of (Ni)-Ni_3Al-NiMo were obtained. The isothermal section of [1999Lu] at 1100 °C is redrawn in Fig. 5. The equilibria near the Al corner (not determined by [1999Lu]) are shown schematically for consistency with the binary data and the results of [2002Gru].

Using the new experimental data in conjunction with the literature data, the CALPHAD method was used by [1999Lu] to optimize the interaction parameters. Two different thermodynamic models were used. Following [1998Hua], the face-centered-cubic (fcc) and bcc phases were modeled as random substitutional solid solutions and the ordered $L1_2$ (Ni_3Al-based) and the ordered $B2$ (NiAl-based) phases were treated as hypothetical disordered phases (different from fcc and bcc), with an ordering contribution to the Gibbs energy. In the second model, following [1997Ans], a single Gibbs energy function was used to describe the disordered fcc and the ordered $L1_2$ or to describe the disordered bcc and $B2$. [1999Lu] pointed out that even though the model of [1997Ans] is more realistic and logical, the two models yielded almost the same results. The experimental data selected for the optimization were those of [1983Wak], [1984Mir], [1989Hon], and [1999Lu]. The tie-line data for the γ/γ' and $\gamma'/B2$ equilibria determined by [1994Jia] were not used in the optimization. The assessed interaction parameters using both models were listed by [1999Lu]. The model of [1998Hua] gave a somewhat better

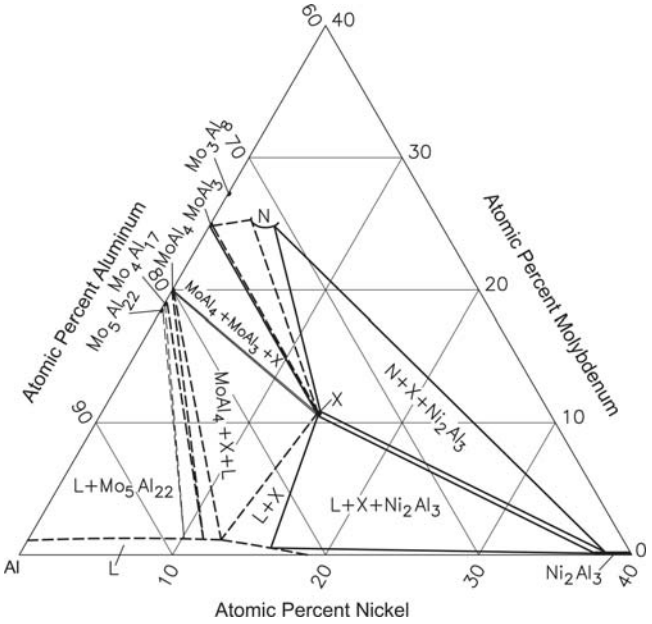

Fig. 3 Al-Mo-Ni partial isothermal section at 950 °C [2002Gru]

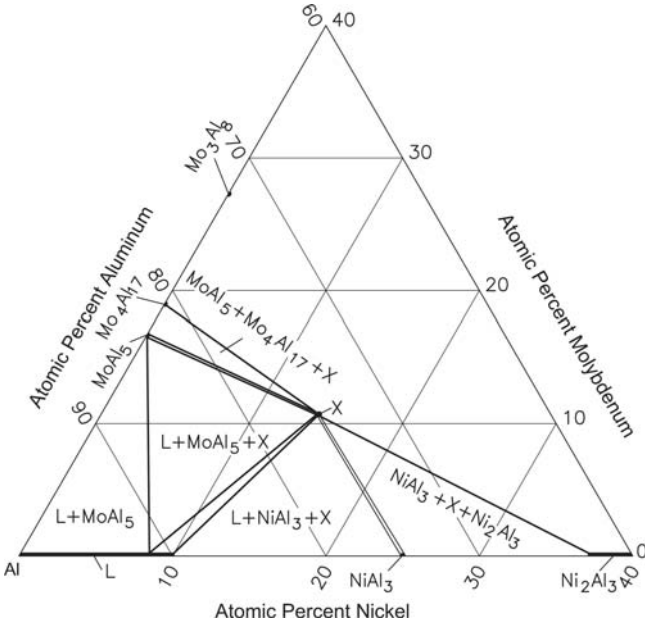

Fig. 4 Al-Mo-Ni partial isothermal section at 800 °C [2002Gru]

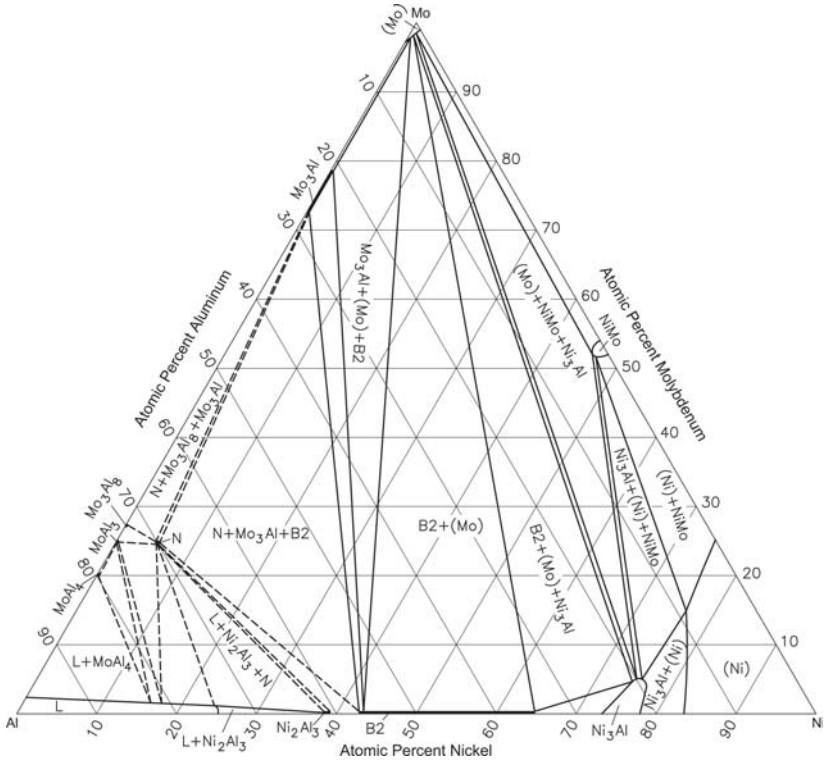

Fig. 5 Al-Mo-Ni isothermal section at 1100 °C [1999Lu]

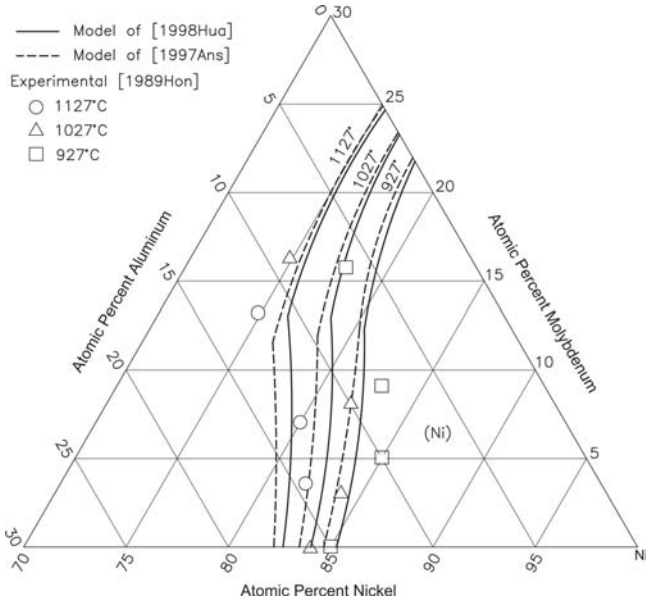

Fig. 6 Al-Mo-Ni computed solvus of (Ni) [1999Lu]

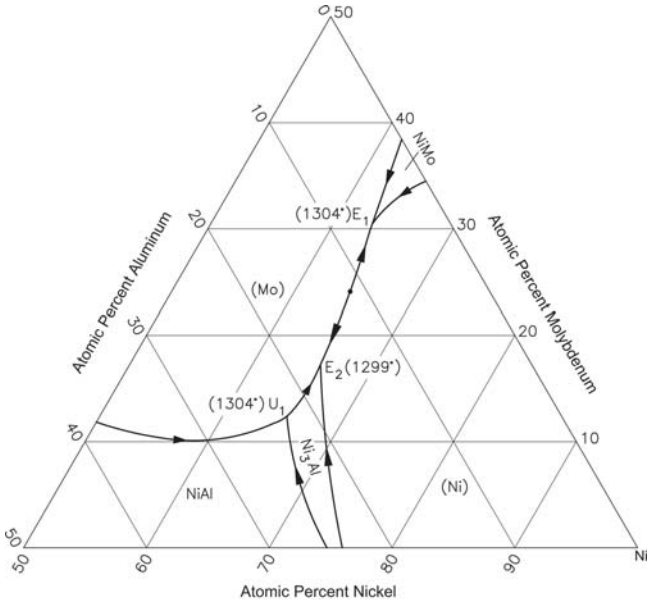

Fig. 7 Al-Mo-Ni computed liquidus projection [1999Lu]

agreement in some cases, as illustrated for the (Ni) solvus in Fig. 6.

Using the parameters derived from the model of [1998Hua], [1999Lu] computed isothermal sections for the Ni-rich region at 1280, 1260, 1200, 1171, 1100, 1093, 1038, 1000, 927, and 880 °C and compared them with the experimental data of [1989Hon], [1988Mas], [1984Mir], [1983Wak], and [1983Nas]. The agreement was generally satisfactory.

[1999Lu] also computed a vertical section at 65 at.% Ni and a liquidus projection for the Ni-rich region. The liquidus projection is redrawn in Fig. 7. [1991Mas] constructed two vertical sections at 14 and 17 at.% Al, using alloys containing up to 24 at.% Mo.

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