Aluminium – Boron – Nickel

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Literature Data

Transient liquid bonding (TLB) using B-Ni eutectic filler material proved to be an efficient process to produce interface-less joints in Ni-base superalloys. Control of process parameters avoiding the appearance of brittle boride phases, however, requires profound understanding of the diffusion process, phase boundaries and ultimately of phase equilibria in the ternary Al-B-Ni system.
Several research groups have dealt with the experimental constitution of the Al-B-Ni ternary system: (a) [1962Sta1, 1963Sta] proposed a tentative liquidus projection for Ni-rich alloys (50-100 at.% Ni) from metallographic and XRD inspection on about 40 ternary alloys prepared by high frequency induction melting in sintered alumina crucibles (air-cooled cast melts). An 800 °C isothermal section was constructed [1962Sta1] based on alloys annealed in evacuated silica capsules for 300 h; (b) [1973Cha] prepared two isothermal sections at 800°C and 1000°C, based on XPD and microstructural analyses. The specimens (elemental powder compacts) were annealed in evacuated silica capsules at 1000, 800 and 600°C for not less than 150, 600 and 1000 h, respectively. Alloys containing more than 60 mass% Al were investigated at 600°C. Results of [1973Cha] confirmed the formation of the τ_1 phase with an extended homogeneity region, but showed the existence of further ternary compounds, τ_2 Ni_5AlB_4 and τ_3 Ni_8AlB_{11}; (c) crystallographic studies were performed of the ternary boride, τ_1, ($Cr_{23}C_6$ type structure) with an extended homogeneity region by [1962Sta2, 1963Sta] and [1998Hil]; (d) a thermodynamic assessment of the Al-B-Ni system is due to [1999Cam] and in refined version from [2000Cam].
Phase relations in the Al-B-Ni system have been reviewed by [1989Sch, 1990Sch, 1999Cam].

Binary Systems

The binary boundary systems Al-Ni and B-Ni are accepted from critical assessments of [2003Sal] and [Mas2], respectively. Although the system Al-B has been assessed and thermodynamically modelled by several authors, the system was adopted in the form described in Al-B-C [2003Gry].

Solid Phases

Three ternary phases, τ_1 to τ_3, were detected by [1973Cha] (Table 1), the latter (Ni_8AlB_{11}) with a polymorphic transition in the temperature range from 800 to 1000°C transforming from a monoclinic high-temperature structure to an unidentified structure at 800°C. The ternary phase, τ_2 Ni_5AlB_4, was only observed in the 800°C isothermal section [1973Cha]; its crystal structure is still unknown.
Whilst both research groups [1973Cha, 1962Sta1, 1962Sta2, 1963Sta] agree on the existence of the τ_1 phase with an extended region of homogeneity, discrepancy exists on the extent of the τ_1 phase at 800°C: [1962Sta1, 1962Sta2, 1963Sta] report on a narrow phase field extending from 21 to 34 at.% boron: $Ni_{20}Al_3B_{6\text{-}12}$ starting from the B-poor composition at 70Ni-8Al-22B (at.%) to the B-rich composition at 60Ni-10Al-30B (at.%). According to [1973Cha] the τ_1 phase region at 800°C is rather small extending at a constant Al-content of 10.3 at.% from 20 to 28 at.% B. The τ_1 phase field is more extensive at 1000°C ranging from 20 to 34 at.% B [1973Cha] in close resemblance to the data listed by [1962Sta1, 1962Sta2, 1963Sta] for 800°C. Only half the number of alloys were examined by [1973Cha] compared with [1962Sta1] but they were annealed for twice the time. Although it is difficult to judge, which version corresponds to true equilibrium, we may assume that Stadelmaier's results have suffered from lack of equilibrium and rather reveal the phase relations at about 1000°C.
[1962Sta2, 1963Sta] tried to explain the unusually large homogeneity region of τ_1 with strongly varying boron content on the basis of a partial replacement of single boron atoms by boron-pairs within the Archimedian Ni- prisms in the crystal structure of the $Cr_{23}C_6$ type.

A detailed structural investigation of the homogeneity region of τ_1 employing X-ray single crystal diffractometry [1998Hil] suggests two substitution mechanisms: (i) Ni/Al replacement $Ni_{20+x}Al_{3-x}B_6$ and (ii) substitution of Al-atoms by boron tetrahedra $Ni_{20}Al_{3-2y}B_{6+8y}$. Whilst the variation of composition from $Ni_{20}Al_3B_6$ to $Ni_{20}Al_{2.4}B_{8.4}$ seems to follow the experimental observations of [1962Sta1] and [1973Cha], the endpoints of the structural series $Ni_{20}Al_{3-2y}B_{6+8y}$ for $0.3 \le y \le 1$ are in conflict with the experimental phase region and certainly provoke further experimental studies.
The possibility of ternary phase formation by reaction of transition metal borides with Al in terms of the electron structure of the transition metals was discussed by [1981Ser].

Invariant Equilibria

Figure 1 presents the reaction scheme for the Ni-rich part of the system. The maximum melting point of congruently-melting τ_1, was said to be at a composition close to 60Ni-9Al-31B (at.%), however, no melting temperature was reported [1962Sta1]. Four ternary invariant reactions were proposed by [1962Sta1], although no temperatures were given and composition of the phases involved can only be guessed from a graph presented; see Table 2. Furthermore the B-Ni binary phase diagram adopted by [1962Sta1] has been revised since, requiring amendments to the reactions proposed by [1962Sta1]: the ternary eutectic reaction $L \rightleftharpoons (Ni)+\tau_1+Ni_3B$, originally had Ni_2B as a product phase, but the transition reaction U_1, $L+NiAl \rightleftharpoons Ni_3Al+\tau_1$ and the ternary eutectic reaction E_1, $L \rightleftharpoons (Ni)+Ni_3Al+\tau_1$, are unchanged. The primary solidification region for "Ni_3B_2" [1962Sta1] needs to be replaced by the phases o-$Ni_4B_3$and m-Ni_4B_3. According to [1962Sta1] there are maximum points on three of the monovariant curves. These correspond to the pseudobinary equilibria $L \rightleftharpoons (Ni)+\tau_1$, $L \rightleftharpoons NiAl+\tau_1$ and $L \rightleftharpoons Ni_3B+\tau_1$.
The fourth reaction, $L \rightleftharpoons \tau_1+NiAl+NiB$, as stated by [1962Sta1], is rather unlikely, as this region of the ternary phase diagram is relatively complex due to binary reactions, which will extend into the ternary in a manner currently unknown.
The thermodynamic calculation [1999Cam] lists the same four ternary invariant reactions as observed by [1962Sta1]. Calculated reaction temperatures and compositions of the liquid phases involved are given dependent on two different assessments for the Al-Ni binary, [1997Ans] and [1998Hua]. There are, however, a series of major discrepancies, as seen from the comparison in Table 2.
A detailed study by DTA and Knudsen-effusion mass spectroscopy [1990Dha] comparing four binary alloys, $Ni_{1-x}Al_x$, $x = 0.20$, 0.23, 0.24 and 0.26 with those containing 0.5 at.% B, showed that minor additions of boron altered considerably the phase boundaries.

Liquidus Surface

A liquidus surface for the region with more than 50 at.% Ni was constructed by [1962Sta1]. As mentioned under "Invariant Equilibria", it needs revision to comply with the accepted binary B-Ni system. The region with less than 30 at.% B is shown in Fig. 2. Due to casting alloys in air (eventual burn up of boron), Stadelmaier's reaction isotherms, E_1, E_2 and U_1 may have shifted to lower boron contents than shown [1962Sta1]. This would be in better agreement to a non-depleted eutectic E_1.

Isothermal Sections

A partial isothermal section at 1000°C (see Fig. 3) was determined by [1973Cha] for the region Ni-NiAl-B. Phase equilibria at 800°C are presented in Fig. 4 and are essentially based on the investigation by [1973Cha]: alloys containing more than 60 at.% Al were annealed at 600°C. However, no complete 600°C isothermal section was reported by [1973Cha]. For an extensive discussion of the homogeneity region of the τ_1 phase, see section "Solid Phases". Both isothermal sections, Fig. 3 and Fig. 4, have been amended to comply with the accepted binary boundary systems. Changes in particular concern the phase AlB_{12} as well as the solubility of Al and Ni in (βB).

Thermodynamics

A thermodynamic modelling of the system Al-B-Ni was performed by [1999Cam] using the CALPHAD approach with Redlich-Kister polynomial description of the Gibbs energy functions [1985Sun]. The binary systems Al-B and B-Ni were included in the modelling. In contrast to earlier treatment, boron was taken as an interstitial element in the solid metal solutions and in Ni_3Al [1999Cam]. Although most of the experimental phase relations are revealed by the thermodynamic calculation, there are three major points of disagreement:
(i) a rather high thermal stability of AlB_{12} favours equilibria with AlB_{12} over those with (βB) i.e. at 800°C rather AlB_{12}+NiB, AlB_{12}+Ni_8AlB_{11}, AlB_{12}+Ni_5AlB_4, than (βB)+Ni_8AlB_{11}, (βB)+Ni_5AlB_4;
(ii) a higher stability is calculated at 1000°C for the two-phase equilibrium m-Ni_4B_3+Ni_8AlB_{11} than the experimentally observed NiB+τ_1;
(iii) the calculated homogeneity region of τ_1 is still significantly smaller than experimentally observed.
With respect to transient liquid bonding, particularly the location of the three-phase field τ_1+(Ni)+liquid at 1200°C and 1300°C was discussed with respect to the variation obtained in the calculation, when two different assessments were used for the Al-Ni binary, namely [1997Ans] and [1998Hua].
Based on the thermodynamic assessment [1999Cam], TLP-simulations and experiments were performed by [2000Cam] monitoring the diffusion path for the TLP bonds in a Ni-10.3Al (at.%) alloy + Ni-10B (at.%) filler material at 1315°C after isothermal holds of 1, 900, 1800 and 3600 sec. Composition-dependent diffusion mobilities were assessed for the ternary system. Adjusting the free energy of the τ_1 phase to yield its experimentally observed mole fraction, the simulations predicted the observed precipitation and later dissolution of the τ_1 phase during the bonding process [2000Cam]. The corresponding calculated tie-triangle at 1315°C was given as: L(87Ni3Al10B)+τ_1(69Ni310Al21B)+(Ni) (92.3Ni6.8Al0.5B) (read from diagram in at.%).

Notes on Materials Properties and Applications

Low density, high oxidation resistance and unusual yield strength dependence on temperature have raised considerable interest in boron-doped polycrystalline intermetallic alloys, NiAl and Ni_3Al, as high temperature structural materials. The acute increase in ductility when a small amount of boron (less than 1 at.%) is added to Ni_3Al has evoked a series of investigations, which are briefly summarized in the following. The microstructure of boron-doped N-atomized Ni_3Al powders $(Ni_{0.76}Al_{0.24})_{99.75}B_{0.25}$ and $(Ni_{0.76}Al_{0.25})_{99}B_1$, has been characterized by LOM and REM revealing remarkable variation in solidification morphology and phase reaction as a function of powder size [1989Hua]: the degree of boron segregation appeared to be significantly reduced, as the formation of $M_{23}B_6$-boride was suppressed in powders below 30 μm diameter.
Ultimate and yield strength at 20 and 800°C of monocrystalline Ni_3Al increase with the B-content to reach a maximum for 0.52 and 1.37 at.% B, dropping slowly for higher B-contents up to 2.22 at.% [1991Guo]. When the B-content exceeds 1.37 at.%B, a eutectic structure Ni_3Al+$Ni_{20}Al_3B_6$ was formed [1991Guo]. Arc melting behavior of continuously cast thin sheet and cast ingots of Ni_3Al containing B and Zr was studied by [1991Li]. The evolution of recrystallization texture in cold rolled $Ni_{76}Al_{24}$(B) on annealing has been investigated by [2000Cho] dividing the annealing process into three stages: recovery, reordering and recrystallization.
Monitoring the distribution of boron in a rapidly solidified alloy, $(Ni_{76}B_{24})_{99.76}B_{0.24}$, via atom probe field-ion microscopy, [1987Hor] found B to segregate to both anti phase boundaries and grain boundaries. A ~0.4 to 1.2 nm thick boron-enriched phase was observed on most of the grain boundaries in a nonuniform distribution. Microstructure and mechanical properties of rapidly solidified Ni-Al ribbons (68-90 at.% Ni) with 0, 200, 2000 and 4000 mass ppm B showed absence of antiphase domain boundaries for hypo-eutectic compositions but a bimodal distribution of APD: for hyper-eutectic compositions [1996Lim]: explanation was based on the metastable Ni-Al phase diagram. Single crystals of Ni_3Al, doped with 0 to 1 at.% B, were examined by TEM after slight compressive deformation. APB and SISF (superlattice intrinsic stacking fault) energies were calculated [1991Yan].

Intergranular segregation of B in Ni_3Al was investigated by [1992Cho] in terms of equilibrium segregation and segregation kinetics on high purity $Ni_{76}Al_{24}$ alloys containing 0.048, 0.144, 0.240 and 0.480 at.% B, aged from 600 to 1050°C to attain equilibrium. The energy of binding of a B-atom to the grain boundary was calculated to be in the range of 0.15-0.45 eV per atom increasing with increasing temperature and decreasing bulk B-content [1992Cho]. The diffusion coefficient of B in Ni_3Al at 500°C was given as $5 \cdot 10^{-21} m^2 \cdot s^{-1}$, at 700°C between 10^{-16} and $10^{-17} m^2 \cdot s^{-1}$. The activation energy for diffusion of B in Ni_3Al was reported to be between 200 and 300 $kJ \cdot mol^{-1}$.

Atomistic simulations (EAM= embedded atom method [2001Zhe, 2002Zhe], and LMTO= Linearized Muffin Tin Orbital method [1990Che] have been employed to study the bulk effects of B on the Ni_3Al -*x*B grain boundary [2001Zhe, 2002Zhe] proposing that B atoms induce Ni for Al substitution and as a consequence B for Ni substitution. It was found that when *x* increases from 0.1 to 1.0 the probability for B to occupy interstitial sites decreases from 97.3 to 38.8 % whilst substitutional occupancy increases from 2.7 to 61.2 % [2001Zhe, 2002Zhe].

Defect strengthening and solution hardening by the addition of boron to NiAl alloys has been studied by [1992Jay, 1993Wu, 1993Tan]. The addition of 0.05 mass% B was found to increase the lattice parameter and the hardness of Ni-rich NiAl, whereas there was no effect for Al-rich NiAl.

The diffusive behavior of B-atoms in Ni_3Al (0.98 at.% B) alloys was investigated by positron annihilation: to achieve moderate segregation of B-atoms to grain boundaries without boride formation, the alloy was recommended to be cooled in air after anneal at high temperature [1995Li].

$Ni_{20}Al_3B_6$ was said to exhibit temperature independent susceptibility down to liquid nitrogen temperatures [1967Hir]

By reacting compacted powders of $1Ni_2B$+5Al at 675°C under argon and subsequently dissolving the Al with 6N KOH, [1966Jah] was able to produce a porous active B-Ni catalyst.

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Table 1: Crystallographic Data of Solid Phases

Phase/ Temperature Range [°C]	Pearson Symbol/ Space Group/ Prototype	Lattice Parameters [pm]	Comments/References
(βAl)	*hP*2 *P6₃/mmc* Mg	*a* = 269.3 *c* = 439.8	at 25°C, 20.5 GPa [Mas2]
(αAl) < 660.452	*cF*4 Fm$\bar{3}$m Cu	*a* = 404.96	at 25°C [Mas2] dissolves 0.01 at.% Ni at 639.9°C [Mas2]
(Ni) < 1455	*cF*4 *Fm$\bar{3}$m* Cu	*a* = 352.40	at 25°C [Mas2] dissolves 20.2 at.% Al at 1385°C [Mas2] dissolves 0.3 at.% B at 1093°C[Mas2]
(βB) < 2092	*hR*333 *R$\bar{3}$m* βB	*a* = 1093.30 *c* = 2382.52 *a* = 1096.5 *c* = 2386.8 *a* = 1096.15 *c* = 2385.44 *a* = 1095.84 *c* = 2385.46	[Mas2, 1993Wer] dissolves up to ca 2 at.%Ni at 1035°C at AlB_{31} [V-C2] for $NiB_{48.5}$ [V-C2] for $NiB_{\sim 20}$ [V-C2]
Al_2B_3 ≤ 525	*hR** Al_2B_3 (?)	*a* = 1840 *c* = 896	at 60 at.% B [1992Var] Metastable?

Phase/ Temperature Range [°C]	Pearson Symbol/ Space Group/ Prototype	Lattice Parameters [pm]	Comments/References
AlB_2	*hP*3		temperature from [2000Hal]
≤ 956±5	*P6/mmm*	*a* = 300.6	[1994Dus]
	AlB_2	*b* = 325.2	
		a = 300.67 ± 0.01	[2002Zhe1]
		b = 325.36 ± 0.02	
		a = 300.43± 0.03	[1999Bur] for $Al_{0.9}B_2$
		b = 325.19 ± 0.06	
αAlB_{12}	*tP*216	*a* = 1015.8	[1994Dus]
≤ 2050	$P4_12_12$	*c* = 1427.0	$\rho_{exp.}$ = 2.65 Mgm^{-3}
	αAlB_{12}	*a* = 1014.93 ± 0.07	[2002Zhe1]
		c = 1425.0 ± 0.5	
γAlB_{12}	*oP*384	*a* = 1014.4	[1994Dus, 2000Hig] metastable phase or
	$P2_12_12_1$	*b* = 1657.3	ternary product stabilized by small
	γAlB_{12}	*c* = 1751.0	amounts of impurity metals present in Al flux grown material; $\rho_{exp.}$= 2.56Mgm^{-3}
$NiAl_3$	*oP*16	*a* = 661.3 ± 0.1	[1996Vik] [Mas2]
< 856	*Pnma*	*b* = 736.7 ± 0.1	
	$NiAl_3$	*c* = 481.1 ± 0.1	
		a = 659.8	[1997Bou, V-C2]
		b = 735.1	
		c = 480.2	
Ni_2Al_3	*hP*5		36.8 to 40.5 at.% Ni [Mas2]
< 1138	$P\bar{3}m1$	*a* = 402.8	[1997Bou, V-C2]
	Ni_2Al_3	*b* = 489.1	
Ni_3Al_4	*cI*112	*a* = 1140.8 ± 0.1	[1989Ell, V-C2]
< 702	$Ia\bar{3}d$		
	Ni_3Ga_4		
NiAl	*cP*2		42 to 69.2 at.% Ni [Mas2]
< 1651	$Pm\bar{3}m$	*a* = 286.0	[1987Kha]
	CsCl	*a* = 287	at 63 at.% Ni [1993Kha]
		a = 288.72 ± 0.02	at 50 at.% Ni [1996Pau]
		a = 287.98 ± 0.02	at 54 at.% Ni [1996Pau]
Ni_5Al_3	*oC*16		63 to 68 at.% Ni [1993Kha, Mas2]
< 723	*Cmmm*	*a* = 753	at 63 at.% Ni [1993Kha]
	Pt_5Ga_3	*b* = 661	
		c = 376	
Ni_3Al	*cP*4		73 to 76 at.% Ni [Mas2]
< 1372	$Pm\bar{3}m$	*a* = 356.77	[1986Hua]
	$AuCu_3$	*a* = 358.9	at 63 at.% Ni [1993Kha]
		a = 356.32	disordered [1998Rav]
		a = 357.92	ordered [1998Rav]
		a = 357.3	for $Ni_{75}Al_{24}B$ [V-C2]

Phase/ Temperature Range [°C]	Pearson Symbol/ Space Group/ Prototype	Lattice Parameters [pm]	Comments/References
Ni_2Al_9	*mP*22 $P2_1/c$ Ni_2Al_9	$a = 868.5 \pm 0.6$ $b = 623.2 \pm 0.4$ $c = 618.5 \pm 0.4$ $\beta = 96.50 \pm 0.05°$	Metastable [1988Li, 1997Poh]
Ni_xAl_{1-x}	*tP*4 *P4/mmm* AuCu	 $a = 383.0$ $c = 320.5$ $a = 379.5$ $c = 325.6$ $a = 379.5$ $c = 325.6$ $a = 375.1$ $c = 330.7$ $a = 379.9$ to 380.4 $c = 322.6$ to 323.3 $a = 371.7$ to 376.8 $c = 335.3$ to 339.9 $a = 378.00$ $c = 328.00$	Martensite, metastable $0.60 < x < 0.68$ [1993Kha] at 62.5 at.% Ni [1991Kim] at 63.5 at.% Ni [1991Kim] at 66.0 at.% Ni [1991Kim] at 64 at.% Ni [1997Pot] at 65 at.% Ni [1997Pot] [1998Sim]
	*m***	$a = 418$ $b = 271$ $c = 1448$ $\alpha = 90°$ $\beta = 93.4°$ $\gamma = 90°$	[1992Mur]
Ni_2Al	*hP*3 $P\bar{3}m1$ CdI_2	 $a = 407$ $c = 499$	Metastable [1993Kha]
	*aP*126 $P\bar{1}$	$a = 1252$ $b = 802$ $c = 1526$ $\alpha = 90°$ $\beta = 109.7°$ $\gamma = 90°$	[1994Mur]
D_1	decagonal		Metastable [1988Li]
D_4	decagonal		Metastable [1988Li]
Ni_3B < 1156	*oP*16 *Pnma* Fe_3C	$a = 521.99$ $b = 661.46$ $c = 436.30$ $a = 530.62$ $b = 667.50$ $c = 444.14$	at 25°C [V-C2, Mas2] at 1100°C [V-C2]

Phase/ Temperature Range [°C]	Pearson Symbol/ Space Group/ Prototype	Lattice Parameters [pm]	Comments/References
Ni_2B < 1125	*tI*12 *I4/mcm* $CuAl_2$	*a* = 499.1 *c* = 424.7	[V-C2, Mas2]
o-Ni_4B_3 < 1025	*oP*28 *Pnma* *o*-Ni_4B_3	*a* = 1195.40 *b* = 298.15 *c* = 656.84	at 41.4 at.% B [V-C2, Mas2]
m-Ni_4B_3 < 1031	*mC*28 *C2/c* *m*-Ni_4B_3	*a* = 642.82 *b* = 487.95 *c* = 781.90 β = 103.47°	at 43.9 at.% B [V-C2, Mas2]
NiB < 1035	*oC*8 *Cmcm* CrB	*a* = 292.9 *b* = 739.2 *c* = 296.1	[V-C2, Mas2]
* τ_1, $Ni_{20}Al_3B_{6+x}$	*cF*116 $Fm\bar{3}m$ $Cr_{23}C_6$	*a* = 1048.5 *a* = 1048 *a* = 1062 *a* = 1049.5 *a* = 1055.2 *a* = 1048.5 9 *a* = 1051.10 *a* = 1051.93 *a* = 1056.89 *a* = 1058.95 *a* = 1059.22 *a* = 1061.67	$Ni_{7.1}Al_{0.8}B_{2.1}$ [V-C2] $Ni_{20}Al_3B_6$ [1973Cha]; 1000°C $Ni_{20}Al_3B_{12}$ [1973Cha]; 1000°C alloy 65Ni10Al25B [1962Sta2]; 800°C# alloy 60Ni10Al30B [1962Sta2]; 800°C# $Ni_{20.5}Al_{2.5}B_6$ [1998Hil] $Ni_{20}Al_3B_6$ [1998Hil] $Ni_{20}Al_{2.7}B_7$ [1998Hil] $Ni_{20}Al_{2.4}B_{8.4}$ [1998Hil] $Ni_{20}Al_{1.1}B_{13.5}$ [1998Hil] $Ni_{20}Al_{1.2}B_{13.2}$ [1998Hil] $Ni_{20}AlB_{14}$ [1998Hil]
$Ni_{23}B_6$		-	metastable, [1997Bat]
* τ_2, Ni_5AlB_4 ≤ 800	-	-	[1973Cha]
* τ_3, Ni_8AlB_{11}(h) > 800	monoclinic	*a* = 3580 *b* = 1093 *c* = 1630 β = 112°	[1973Cha], from single crystal rotation photographs
* τ_3, Ni_8AlB_{11}(r) ≤ 800	-	-	[1973Cha]

for a discussion of phase region and equilibrium conditions, see section “Solid Phases”.

Table 2: Invariant Equilibria (data are only available for the liquid phase)

Reaction	T [°C] Experimental, [1962Sta1]	T [°C] Calculated,* [1999Cam]	Type	Phase	Composition (at.%) Experimental#			Composition (at.%) Calculated [1999Cam]		
					Ni	Al	B	Ni	Al	B
$L \rightleftharpoons NiAl+\tau_1$	-	-	e, max	L	-	-	-	-	-	-
$L+NiAl \rightleftharpoons Ni_3Al+\tau_1$	<1369	1348	U_1	L	~69	~13	~18	73.8	24.1	2.1
$L \rightleftharpoons (Ni)+\tau_1$	-	-	e, max	L	~75	~7	~18	-	-	-
$L \rightleftharpoons (Ni)+Ni_3Al+\tau_1$	$<U_1$	1349	E_1	L	~69	~10	~18	75.4	22.5	2.1
$L \rightleftharpoons (Ni)+Ni_3B+\tau_1$	<1093	1090	E_2	L	~79	~3	~18	83.8	0.03	16.2
$L \rightleftharpoons NiAl+NiB+\tau_1$	-	692	E_3	L	~52	~10	~38	34.2	17.9	47.9

* calculated data refer to the Al-Ni binary modeled by [1998Hua]; data changed slightly if the model of [1997Ans] was used; for details see original work by [1999Cam].

\# values read from small diagram [1962Sta1].

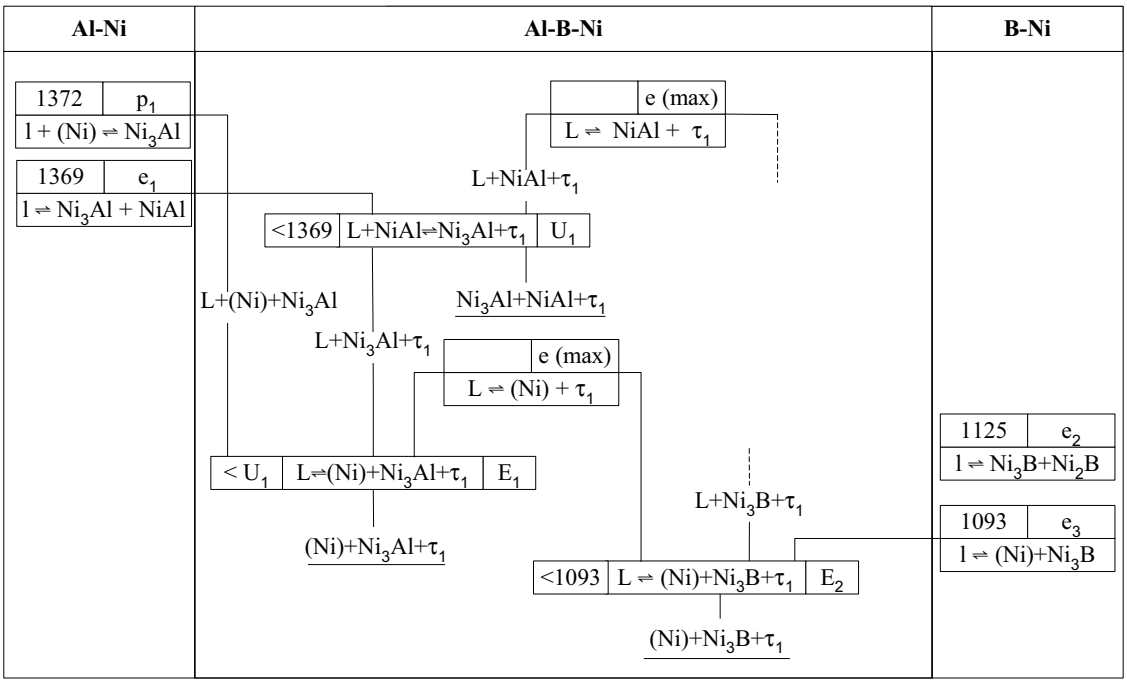

Fig. 1: Al-B-Ni. Partial reaction scheme

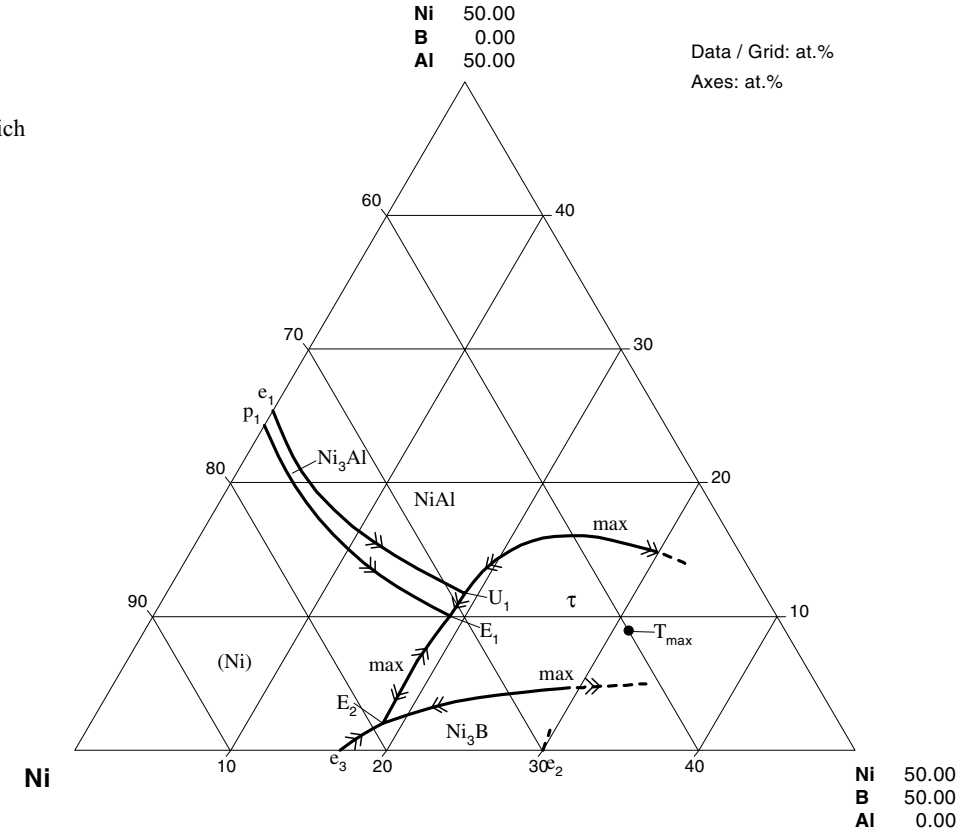

Fig. 2: Al-B-Ni
Partial liquidus surface in the Ni-rich region after [1962Sta1]

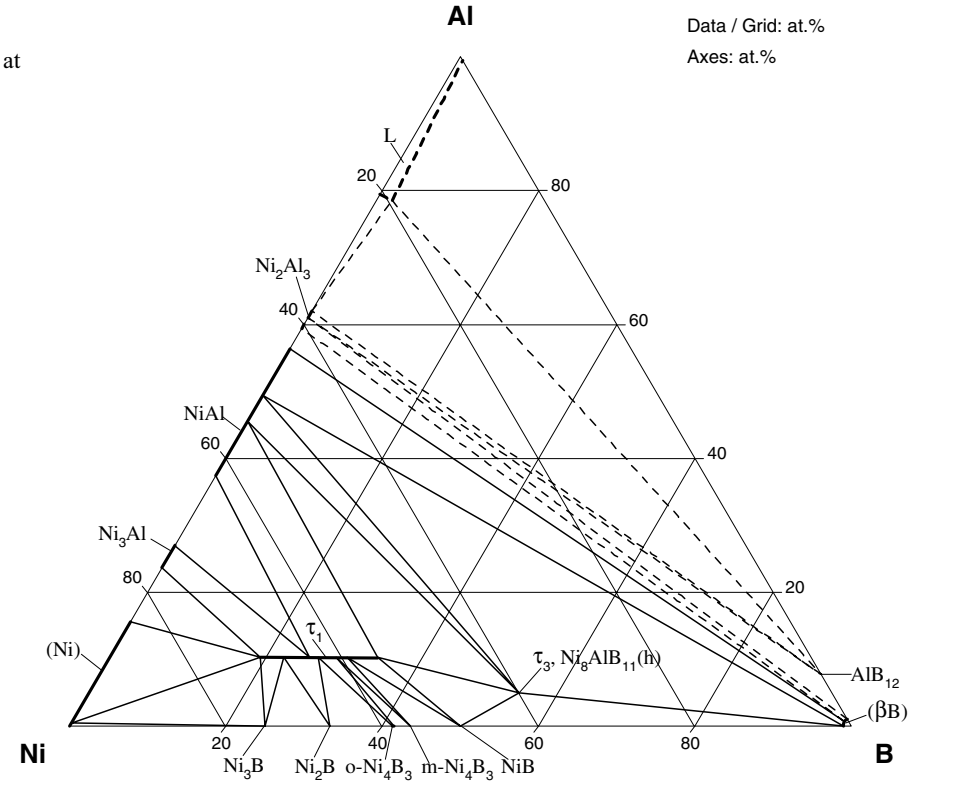

Fig. 3: Al-B-Ni.
Isothermal section at 1000°C

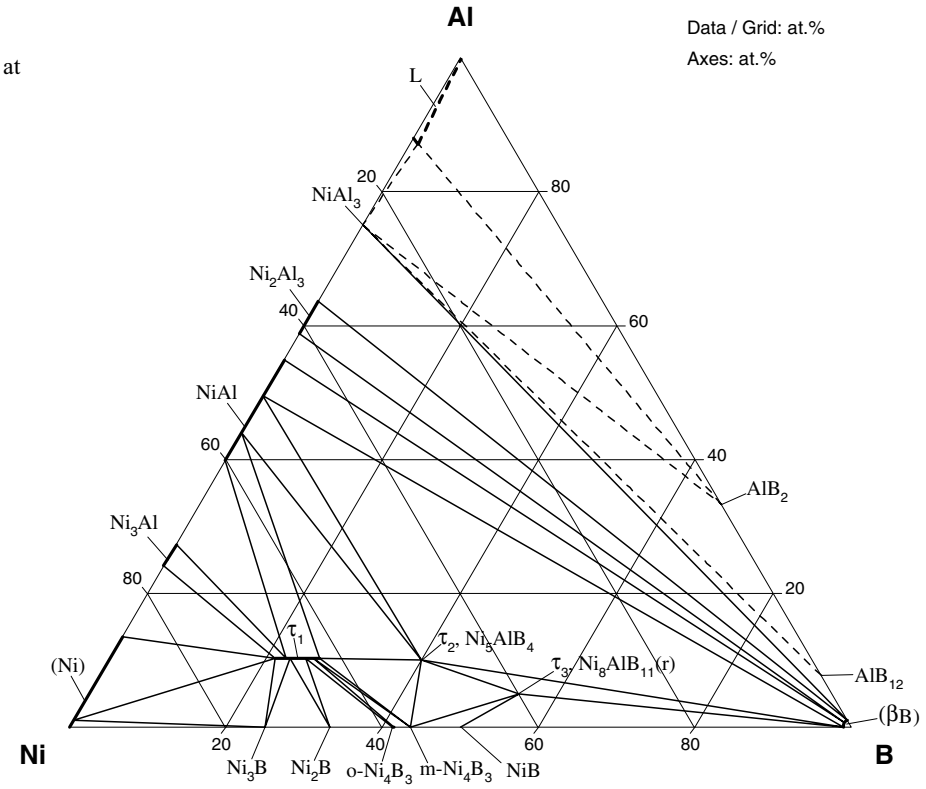

Fig. 4: Al-B-Ni.
Isothermal section at 800°C