

Aluminium – Chromium – Nickel

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Literature Data

The Al–Cr–Ni system has been investigated in many works, however Al–Cr–Ni equilibrium diagram has not been determined in the whole composition range so far. After assessments of [1984Mer1] and [1991Rog], in accordance with which there are no ternary phases in the system, several reports with data on the ternary phases in Al-rich part of phase diagram have appeared. The major part of the earlier works had been dedicated to the Cr–Ni edge of the system. The present evaluation updates work of [1991Rog] and considers all data available.

Besides the early cursory study [1933Roe], summarized by [1943Mon], dealing with the influence of Cr and Ni additions on the (Al) solid solution, the first detailed investigation of the NiAl–Ni–Cr₅₀Ni₅₀ subsystem, including a series of isothermal sections at 750, 850, 1000 and 1150°C as well as tentative mapping of the fields of primary crystallization, was done by [1952Tay2]. The alloys, usually weighing 50 g, were prepared by induction melting of the high purity constituting elements (impurities not given) in alumina-lined crucibles under a low pressure of hydrogen. The alloys were studied in the as-cast state as well as after annealing at various temperatures: 750°C (500 h), 850°C (500 h), 1000°C (100 h) and 1150°C (48 h) employing X-ray powder techniques on filings along with optical microscopy for the alloys cooled from 750, 850°C. Phase relations at higher temperatures were derived by microoptical analysis only. A homogenizing step at 1150 to 1250°C for 100 h was introduced prior to the final annealing procedure. Further studies of partial isothermal sections and reaction isotherms were performed by [1958Bag1] and [1958Bag2] (1150, 1200°C), and by [1982Tu, 1983Lan, 1983Ofo, 1984Car, 1985Ofo, 1987Ofo] and [1989Hon] in the temperature range from 950 to 1200°C. Whereas the latter group of authors mainly employed DTA, EMPA, SEM-EDX techniques to derive precise data for the location of the phase boundaries, diffusion experiments were used by [1984Car, 1987Nes1, 1987Nes2, 1987Nes3] and [1984Mer2] and the work of [1983Ofo, 1985Ofo] and [1987Ofo] involved Knudsen cell mass spectroscopy. The alloys of [1987Nes1, 1987Nes2, 1987Nes3] and [1989Hon] were prepared by arc melting the high purity elements under argon followed by subsequent annealing for 140 h at 1250°C in evacuated silica tubes [1989Hon] or in molybdenum cans [1987Nes1, 1987Nes2, 1987Nes3] with final quenching in water. The diffusion couples, used by [1987Nes1, 1987Nes2] and [1987Nes3] were fabricated from flat polished discs cut from the pre-alloyed and homogenized samples and were further heated in Mo containers at either 1100 or 1200°C for 100 h. A second set of diffusion-bounded couples was prepared by hot-pressing in a stainless steel mixture at 1100°C in vacuum for 1 h under 20 MNm⁻² pressure. These couples were then sealed in evacuated silica capsules, backfilled with Ar, and annealed at 1095 and 1205°C for 100, 300 and 500 h. A similar preparation technique was adopted by [1984Mer2]. The diffusion couples were then sectioned, polished and examined by EMPA. [1983Ofo, 1985Ofo] and [1987Ofo] studied 16 binary and 9 ternary alloys by means of X-ray diffraction, microscopic analysis, Knudsen cell mass spectroscopy and EMPA. The alloys were prepared by melting high purity metals in a vacuum furnace under argon, keeping the melt for some time at certain temperature, followed by chill casting into ingots of 2.5 cm diameter and 15 cm height. For heat treatment, the samples were placed inside an Al₂O₃ tube contained in a sealed evacuated silica tube and annealed for 100 to 200 h; Cr-rich alloys were annealed for 400 h. The more ductile samples were given a 50% hot compression and heat treated at 1250°C for 240 h with 100 extra hours at 1150°C. All samples used in the 1150°C isothermal section were quenched and their composition was checked by chemical analysis. The use of phase diagram information in the understanding of the multi-element diffusion and the correlation of coating/substrate compositions, microstructures and behavior has been demonstrated by [1977Jac1] on Al–Cr–Ni coatings of superalloys prepared by pack aluminization, and a series of isothermal sections around 1000°C have been proposed based on X-ray diffraction, microscopical, chemical and microprobe analysis of the aluminized structures. A large number of experimental investigations are concerned with the constitution of alloys on and in vicinity of the section NiAl–Cr

[1953Kor, 1952Tay2, 1958Bag1, 1958Bag2, 1992Gor] and [1993Cot1], which is not quasibinary if the reported data on directions of tie-lines in (α Cr)+(β NiAl) field [1982Tu, 1985Ofo] are taken into account, and the other results generalized in the thermodynamic assessment [2001Dup]. This conclusion agrees with [2001Com] who reported that the maximum temperature on invariant line $L \rightleftharpoons \alpha + \beta$ [2001Com] was shifted to Al corner. Liquidus surface projection of the NiAl–Cr–Ni subsystem was presented by [1991Rog] having generalized the data of [1952Tay2, 1954Kor, 1955Kor2, 1958Kor]. Using the elemental metals and 25 or 54.2 at.% Al–Ni master alloys of various purities (99.96 mass% Cr or 97 mass% Cr containing 2.5 mass% Al, 0.3 mass% Si and 0.5 mass% Fe), the alloys (50 g each) were prepared by high frequency melting in alumina crucibles either under a $48\text{Al}_2\text{O}_3$ –7MgO–45CaO flux [1954Kor] and [1955Kor2] or under argon [1958Bag1]. The microstructure, hardness and temperature coefficient of resistivity were investigated after quenching from 1200°C in water and after slow cooling; thermal expansion and heat resistance were studied on 1200°C annealed alloys. In the range of compositions near NiAl–Cr section (with chromium content up to 5 at.%) liquidus and solidus surfaces were plotted and phase composition of alloys slowly cooled from 1400°C was presented by [1993Cot1]. [1993Cot1] produced the alloys using nonconsumable arc melting of constituent elements (99.999 mass% Al, 99.99 mass% Cr, 99.95 mass% Ni) in a water-cooled copper crucible under argon gas. Then homogenization was carried out in commercial purity argon gas for 24 h at 1400°C, using heating and cooling rates of 0.083 K/s. DTA was employed using samples prepared by pulverizing ~10 to 50 g of each homogenized casting to powder. Investigation of the NiAl–Cr eutectic alloys was conducted by [1978Vol], and also by [1992Gor] in the composition range 12 to 33 at.% Al. [1992Gor] investigated also eutectic alloys α Cr– γ Ni in the range of 0 to 15 at.% Al. Directional eutectic crystallization was performed by [1970Cli, 1970Wal, 1971Cli, 1973Wal, 1978Vol, 1986Gor] and [1996Ros] on NiAl–Cr alloys produced in an arc furnace, cast into rods in an induction furnace, followed by directional crystallization by the Bridgman method in a vertical furnace. The phases with crystal structure different from one of the binary phases were revealed and investigated by [1982Tu, 1989Zho, 1990Col, 1996Ros, 1997Li, 1997Sat, 2000Uch] and [2001Com]. For the first time the existence of a ternary phase in Al–Cr–Ni has been reported by [1982Tu]. The phase was found in a sample pack-aluminized at 900°C and then it was prepared as single phase (according to metallography) of $\text{Cr}_{15}\text{Ni}_{10}\text{Al}_{75}$ composition. The structure was not determined. A ternary phase with the same composition and four more ones were obtained by [1996Ros] in as-cast and annealed Al-rich alloys. The aspect of the phase diagram concerned and investigated to liquid-solid equilibria in the subsection AlCr–AlNi–Al was investigated by [1996Ros] and [2001Com]. The alloys by both [1996Ros] and [2001Com] were prepared by melting of high purity (99.9 mass%) constituent Al, Cr and Ni. But [1996Ros] carried out the melting process in an inductive cold crucible under an argon atmosphere as well as by the addition of chromium and aluminium to prealloyed NiAl_3 or Ni_2Al_3 compounds, just as [2001Com] used arc melting under an argon atmosphere. The alloys were investigated in as-cast state [2001Com] or in addition after quenching from selected temperature in the DTA experiment process. Compositions of the phases were determined using SEM-EDX [1996Ros, 2001Com], EDS (Link AN10000 system) and EMPA (on a Cameca SX50 microprobe) [1996Ros]. Phase structures were identified by TEM [1996Ros]. The transformation temperatures in alloys were measured by DTA on samples of about 1 g masses using the cooling and heating rates $5\text{K}\cdot\text{min}^{-1}$ or less, however, the fields of primary crystallization are given without direction of temperature change. This was also shown on liquidus surface projection of the AlCr–AlNi–Al subsystem by [2001Com], where the temperatures of the invariant equilibria are also not indicated.

The crystal structure of the ternary phases obtained by rapid solidification and in conditions of conventional cooling rate were investigated by [1989Zho, 1997Li, 1997Sat], and [2000Uch] using SEM-EDX, TEM, and HREM, and modeling. Accurate determination of lattice parameters was used by a series of investigators for a precise characterization of the extent of the mutual solid solution behavior in the Al–Cr–Ni ternary [1952Tay2, 1955Kor1, 1955Kor2, 1958Bag2, 1959Gua, 1960Kor, 1961Ham, 1961Pry, 1963Arb, 1964Ram, 1965Ram, 1971Cli, 1972Sch, 1976Low, 1983Och, 1984Och1, 1984Och2, 1985Mis, 1996Ros, 1997Kre, 1999Zak, 1999Zya] and [2001Com]. Samples were generally prepared by melting under argon followed by various heat treatments under argon in sealed silica capsules. [1960Kor] and [1961Pry] annealed their alloys for 200 h at 1200°C and 100 h at 1000°C in evacuated silica tubes followed by a water quench. Samples of [1961Ham] were prepared by high frequency melting in vacuum, then homogenized for

4 h at 1260°C and hot rolled whereby the material was reheated five times to 1260°C for 10 min. These specimens were then aged at selected temperatures for 12 to 24 h and quenched in iced brine. Whereas [1952Tay2, 1955Kor1, 1955Kor2, 1958Bag2, 1960Kor, 1964Ram, 1965Ram, 1971Cli, 1972Sch, 1976Low, 1984Och1, 1984Och2, 1985Mis] and [1999Tia, 2002Fis] deal with the extent of the β NiAl and γ (Ni) phase regions, the problem of Cr substitution in γ' Ni₃Al is treated by [1952Tay2, 1959Gua, 1961Ham, 1963Arb, 1966Arb, 1984Och1, 1984Och2] and [1985Mis]. In search for a MgCu₂ type phase, [1964Ram] and [1965Ram] studied two alloys: Cr₂₅Ni₁₃Al₆ (arc melted, as-cast) which was a homogeneous Cr₅Al₈ type, and Cr₃₃Ni₃₃Al₃₄ which, as-cast and/or after 85 h at 700°C, revealed a CsCl type phase in accordance with the 750°C isothermal section of [1952Tay2]. Based on earlier investigations by [1959Gua, 1963Arb] and [1966Arb], the extent of the γ' Ni₃Al phase field on alloying with Cr as well as the mode of atom substitution has been investigated by [1983Och, 1984Och1, 1984Och2] and [1985Mis] by means of X-ray diffractometry and metallography, assisted by hardness measurements from [1959Gua] and [1963Arb]. Three series of alloys, Cr_xNi₃Al_{1-x}, Cr_xNi_{3-x}Al and Cr_x(Cr_yNi_{1-y})₃Al_{1-x}, were prepared by [1983Och, 1984Och1, 1984Och2] and [1985Mis], who essentially confirmed earlier data [1966Arb] with Cr replacing Al in Cr_xNi₃Al_{1-x} while in general Cr was found to substitute for both lattice sites in Cr_x(Cr_yNi_{1-y})₃Al_{1-x}.

Binary Systems

The Cr–Ni system is adopted according [Mas2]. The Al–Cr and Al–Ni systems are accepted from [2003Cor] and [2003Sal], respectively.

Solid Phases

Data on the solid phases are presented in Table 1. For the Al-poor range of compositions below the Cr₅Al₃–Ni₂Al₃ tie-line, no ternary phases are found. In the Al-rich corner several ternary phases have been found in as-cast (or annealed and cooled at conventional rate) alloys. Also a hexagonal metastable phase in the rapidly solidified Cr_{0.5}Ni_{0.5}Al₆ alloy together with decagonal and icosahedral quasicrystals was found by [1989Zho]. It appears during the quasicrystal-to-crystal transition and its structure closely resembles that of the decagonal quasicrystal and stable orthorhombic phases because of the presence of icosahedral subunits. The hexagonal phase after heating up to 300°C transforms into an orthorhombic structure. Monoclinic τ_1 , orthorhombic τ_2 , and hexagonal τ_3 , τ_4 , τ_5 phases reported by [1996Ros] and [1997Li, 1997Sat] are of own crystal structure types. The structure of τ_1 and τ_2 are rather complicated, each of them exhibits very large cell parameters and different polytypes and superordering. The high temperature rhombohedral τ_3 and hexagonal τ_4 phases are found to be of related structure. The orthorhombic τ_1' polytype forms because of a periodic multiple (100) twinning of τ_1 . The orientational relationships between the τ_1 monoclinic and its orthorhombic polytype structure are as follows:

$$(010)_{\text{monocl.}} \parallel (001)_{\text{orthor.}}$$

$$(001)_{\text{monocl.}} \parallel (010)_{\text{orthor.}}$$

$$(100)_{\text{monocl.}} \parallel (100)_{\text{orthor.}}$$

and relation between the cell parameters is: $a_{\text{monocl.}} = (c_{\text{orthor.}} \cdot \sin^{-1}\beta)/2$, $b_{\text{monocl.}} = a_{\text{orthorh.}}$, $c_{\text{monocl.}} = b_{\text{orthor.}}$. The a and b parameters of τ_2 appear to correspond to those of the τ_1' phase, which is an orthorhombic polytype of τ_1 . This is confirmed by the orientational relationships between the orthorhombic τ_1 and orthorhombic τ_2 phases. They are as follows:

$$(100)_{\tau_1} \parallel (001)_{\tau_2}$$

$$(001)_{\tau_1} \parallel (100)_{\tau_2}$$

$$(010)_{\tau_1} \parallel (010)_{\tau_2}$$

and relation between the cell parameters is:

$$a_{\tau_2} = b_{\tau_1}, b_{\tau_2} = a_{\tau_1}, c_{\tau_2} = (7a_{\tau_1} \cdot \sin\beta_{\tau_1})/6.$$

Triclinic superstructure of τ_2 (τ_2') is supposed by [1996Ros], with cell parameters being equal:

$a_{\text{tricl.}} = 2a_{\tau_2}$, $b_{\text{tricl.}} = b_{\tau_2}$, $c_{\text{tricl.}} = 2c_{\tau_2}$, or orthorhombic polytype with cell parameters two times larger than those of τ_2 . The orientational relationships between the two structures of τ_3 and τ_4 are as follows:

(III) \parallel (0001)

($\bar{1}\bar{1}0$) \parallel ($\bar{1}2\bar{1}0$)

(II $\bar{2}$) \parallel (I0 $\bar{1}0$) and $\tau_4 = 2a_{\tau_3, \text{rhomb.}} \cdot \sin(\alpha/2)$. These two phases (τ_3 and τ_4) are considered to be high temperature ones. The τ_5 phase is reported as stable at low temperature, and forming through reaction between τ_3 and τ_4 . The structure of the τ_3 , τ_4 and τ_5 phases is very close to that of μCrAl_4 . However substitution of Al or Cr atoms by Ni atoms in the μCrAl_4 structure does not occur or is very small. [2001Com] claimed that they confirmed the phases described by [1996Ros] with the microstructure and the composition measurements. But there are some dissensions in the results of [1996Ros] and [2001Com] concerning phase compositions. The κ phases considered by [1997Li, 1997Sat] and [2000Uch] are the same as τ_4 (h) of [1996Ros]. The latter is shown to be related to icosahedral quasicrystals, and to have a large unit cell with 222 atoms, with transition metals Cr and Ni co-occupying the same positions [1997Li, 1997Sat]. [2000Uch] proposed a new interpretation of this phase structure as modulated one instead of generally accepted description in terms of the aggregations of clusters. The proposed modulated structure is basically composed of close-packed layers with ordered atomic vacancies owing to the occurrence of charge-density waves.

Pseudobinary Systems

There is no pseudobinary section in the Al–Cr–Ni system. But the pseudobinary eutectic exists within an isopleth shifted to the Al corner and its temperature is somewhat higher compared with the accepted one by [1984Mer1, 1991Rog] for the radial Cr – $\text{Ni}_{0.5}\text{Al}_{0.5}$ section. This displacement may be attributed to the experimental observation of [2001Com] and was shown in thermodynamic assessment of [2001Dup]. There is no special investigation concerning the location of the tie-line with maximal temperature in the two-phase (αCr)+ $\text{NiAl}_{1\pm x}$ field.

Invariant Equilibria

Seven ternary invariant equilibria in ternary system are reported (Table 2): maximum decomposition of the liquid e_1 at temperature of about 1445°C (see section Pseudobinary Systems) within maximal tie-line Cr– NiAl_{1+x} ; eutectic decomposition of the liquid E_1 at 1300±20°C, which is the average of the data of 1320±10°C [1952Tay2], 1300°C reported by [1955Kor2], and the calculated ternary eutectic at 1288°C by [1981Sto] as well as at 1280°C by [2001Dup]; a transition type reaction U_1 at 1350°C [2001Dup] which is in a reasonable agreement with value of 1340 ± 10°C reported by [1952Tay2] which, due to the mode of formation of the γ' phase, was slightly altered by [1984Hil, 1988Bre], and two more transition reactions, U_2 and U_3 , involving the binary $\delta\text{Ni}_2\text{Al}_3$ and $\zeta\text{Cr}_5\text{Al}_8$ phases. These reactions were already postulated by [1984Mer2] in the proposed partial reaction scheme for the Al–Cr–Ni system. The solid state transition reaction U_4 at 996°C after [2001Dup] is in a good agreement with a value of 990 ± 3°C [1982Tu], which was earlier proposed at 1000°C by [1952Tay2]. The eutectic four-phase invariant reaction E_2 exists at 634°C [1996Ros], and one of the preceding processes of crystallization, with participation of the θ phase and aluminium, being incongruent in the binary Al–Cr system, $\text{L}+\theta\rightleftharpoons(\text{Al})$, in the ternary system changes the character to congruent one: $\text{L}\rightleftharpoons(\text{Al})+\theta$. The compositions of the phases in U_1 , E_1 , and U_4 equilibria are given by [2001Dup]. The compositions of the other liquid phases are given explicitly by [1952Tay2] and [1955Kor2], and those of the solid phases are constructed by extrapolation from EMPA data given by various research teams [1991Rog]. A partial reaction scheme is presented in Fig. 1.

Liquidus Surface

The liquidus projection, as shown in Fig. 2, is essentially based on the early data by [1952Tay2, 1954Kor, 1958Bag1] and [1993Cot1] followed by critical discussion by [1982Wes, 1984Mer1] and [1991Rog]. Additionally to the isotherms corrected slightly by [1991Rog] with respect to the accepted boundary systems and the melting point of Cr at 1863°C the isotherms based on [2001Dup, 1993Cot1] data are proposed in Fig. 2. Figures 3 and 4 present calculated liquidus surface by [2001Dup] and the β , NiAl phase liquidus surface in vicinity of $\text{Ni}_{0.5}\text{Al}_{0.5}$ composition after [1993Cot1]. The liquidus surface projection in

the Al-rich corner is not established. The experimental data of [1996Ros] and [2001Com] concerning the map of the liquidus surface in the Al-rich corner disagree with each other, although the phases in the as-cast alloys were the same according to [2001Com]. Neither the direction of the monovariant temperatures nor the invariant temperatures are given by [1996Ros]. And the relative dislocation of the primary fields and ternary phase compositions given by [2001Com] are inconsistent with each other from the view point of geometrical thermodynamics and with the shown temperature change direction along the monovariant lines.

Isothermal Sections

The earlier published information [1952Tay2, 1958Bag1, 1958Bag2, 1958Kor, 1982Tu, 1984Mer2, 1983Lan, 1983Och, 1985Ofo, 1987Nes1, 1987Nes2] and [1989Hon], assessed by [1991Rog], and [1992Gor, 1993Cot1, 1994Jia, 1999Tia] is concerned with the NiAl–Cr–Ni subsystem covering the temperature range from $\sim 600^\circ\text{C}$ to solidus. [2001Dup] thermodynamic assessment covered many of the mentioned and recent publications [1990Dav, 1990Kek, 1994Yeu, 1996Wu, 1999Sun]. Generally a good agreement between calculated and experimental data for investigated temperature and composition ranges was obtained. Both, experimental and calculated isothermal sections are presented in Figs. 5 to 16. Figure 17 demonstrates β phase solidus surface in the vicinity of its equiatomic composition. The main characteristic is the increase in the Cr solubility in the γ' phase field with decreasing temperature. Phase relations at 750 and 850°C in the Ni-rich corner for more than 50 at.% Ni are due to [1952Tay2] and presented in Fig. 15 and Fig. 13 with minor adaptation to the accepted binaries [1991Rog]. Isoparametric lines for the size of the unit cell within the γ and γ' phase fields reveal the difficulties in the X-ray resolution of the two bcc-phases in a certain restricted composition range, where the exact overlap of the X-ray high angle reflections tends to mask the duplex structure. Whereas fcc-fcc precipitation usually occurs on the (111) planes, [1952Tay2], in the region with equal lattice parameters of the γ and γ' phases, a preference of the γ' -precipitates for the rectangular-cubic to the usual spheroidal shape growing on the (100) planes is observed. Phase relations, as found in slowly-cooled samples, were reported to essentially correspond to the 750°C isothermal section [1952Tay2]. [1952Tay2] was first to observe the $\beta+\gamma\rightleftharpoons\alpha+\gamma'$ transition reaction, which was attributed to the 1000°C isothermal section. Later thermal analysis by [1982Tu] gave a value of $990 \pm 3^\circ\text{C}$ for the four-phase reaction isotherm. The effects of alloying on the temperature of the $\beta+\gamma\rightleftharpoons\alpha+\gamma'$ transition reaction were studied by [1976Jac]. Employing EMPA, [1982Tu] provided a phase triangulation in the ternary up to 60 at.% Al at 1025°C , which is presented in Fig. 10a. The tip of the γ' phase field has been accurately re-determined by [1983Lan] overcoming the matrix effects of [1982Tu] in the EMPA analysis of the γ' precipitates. Some controversies exist between the phase relations at temperatures around 1025°C according to [1982Tu, 1983Lan] and the isothermal sections designed by [1977Jac1] and [1977Jac2] for the temperature range ~ 950 to $\sim 1100^\circ\text{C}$. The discrepancies mainly concern the postulated $\alpha+\delta$ equilibrium, which is in contradiction to the findings of [1982Tu, 1983Lan] and [1985Ofo]. A second argument concerns the curvature of the α phase boundary in the ternary, which in the original paper by [1977Jac1] was shown with a negative curvature with respect to the Al–Cr boundary, but was later corrected by [1977Jac2] to be consistent with the experimental findings of [1958Bag1, 1982Tu, 1983Lan] and [1985Ofo]. Figure 7a gives the phase relation at 1150°C as determined by [1952Tay2] and [1958Bag1] and by a more complete and herein accepted EMPA study by [1985Ofo]: inconsistencies mainly concern the extensions and widths of the β , γ and γ' phase fields. Whereas [1952Tay2] arrived at a rather exaggerated Cr-solubility in the β phase, the data by [1985Ofo] seem to support earlier results by [1958Bag1] revealing a maximum solubility of 10 at.% Cr at 30 at.% Al. Agreements exist on the maximum Cr solubility in the γ' phase of 8 at.% Cr at 22 at.% Al [1952Tay2, 1958Bag1] and [1985Ofo]. The form and extension of the ζ field [1985Ofo] (Fig. 7a), however, need further investigation. Unlike [1985Ofo], [2001Com] in as-cast alloys found the content of Ni in ζ phase to be equal 3.3 at.% only and Ni decreases slightly the content of Al in ζ . This finding agrees with [1982Tu, 1983Lan] data for 1025°C (Fig. 10a).

Replotting the lattice parameter values obtained by [1952Tay2], [1955Kor1] showed a partially linear variation of the lattice parameters of the $\gamma(\text{Cr, Ni, Al})$ solution at 20 at.% Cr up to 10 at.% Al. Similarly, based on X-ray and metallographic analysis, hardness measurements and electrical resistivity data, [1960Kor] and [1961Pry] showed a linear variation of the lattice parameters of the $\gamma(\text{Cr, Ni, Al})$ solid

solution at 11.1 at.% (10 mass%) Cr up to 16.9 at.% Al (at 1200°C) and up to 13.2 at.% Al (at 1000°C), respectively. The γ/γ' phase boundary as a function of temperature, as assessed by [1961Ham] from extensive metallographic studies by electron microscopy and high temperature X-ray diffractometry in the range from 750 to 1200°C and within the vertical section Ni_3Al – CrNi_3 , reveals slight deviations from the values obtained by [1952Tay2]. The γ/γ' solvus boundary [1961Ham] tends to show larger Al solubility in the γ phase at lower temperatures, but smaller Al contents at temperatures above 1000°C. The data of [1961Ham] are, however, in a good agreement with the EMPA studies of [1982Tu, 1983Lan, 1984Car] and [1985Ofo]. The γ/γ' -solvus lines at 827, 927, 1027 and 1127 were obtained by employing DTA with support from energy dispersive X-ray spectroscopy combined with SEM-ELY. Their results, on the contrary, show slightly smaller Al solubilities in the γ phase with respect to [1961Ham]. Therefore, the data of [1952Tay2] are accepted for the 750 and 850°C isothermal sections, whereas the outline of the γ field at 1025 and 1150°C is accepted according to the measurements by [1982Tu, 1983Lan, 1984Car] and [1985Ofo], despite the fact that their curvature of the Al-rich γ/γ' boundary is opposite to [1989Hon]. Isothermal sections above 1150°C are primarily concerned with the $\alpha+\beta+\gamma$ and the $\beta+\gamma+\gamma'$ equilibria [1958Bag1 and 1958Kor], and [1987Nes1], and [1987Nes2]. Solubility data [1958Bag1] and [1958Kor] are listed in Table 2. An estimated phase field distribution at 1200°C is given by [1987Nes1]. His investigation, however, mainly concerns the $\gamma/\beta+\gamma$ boundary ranging from 57 at.% Ni at 10 at.% Al to 72 at.% Ni at 15 at.% Al.

The calculated by [2001Dup] partial isothermal sections at 750, 800, 850, 900, 1000, 1025, 1100, 1150, 1200 and 1300°C are presented in Figs. 5 to 15. As one can see, the calculation showed a smaller solubility of Cr in γ' and β phases compared with the experimental data [1952Tay1] and [1952Tay2] at 750 and 850°C, and uncommon form of β phase solvus as well as a negative curvature of Cr solvus isotherm at 1150°C. The calculated smaller solubility of Cr in γ' at temperatures lower than 1000°C was confirmed in experiments of [1983Och], who dealt with longer heat treatment time of alloys at 1000°C and their higher purity than [1952Tay2]. These data well agree with calculated isothermal section. Elevated solubility value reported in earlier works could be due to not fully equilibrated samples or due to the poor sensitivity of the experimental methods. The minimal solubility of Cr in the equiatomic βNiAl phase according to the calculation [2001Dup] shown in Figs. 7b, 10b, 13b, 15b is supported by experimental points of [1982Tu] in the isothermal section at 1025°C (Fig. 10a), although approximation made by [1983Lan] does not cover these experimental points.

[1999Tia] found by TEM that the solubility of Cr in βNiAl at Al contents of 48 to 50 at.% is not more than 8 at.% for alloys annealed for 74 h at 1290°C and quenched; and it is less than 2 at.% in the range of 700 to 800°C. These data were obtained on alloys of high purity: starting materials were 99.9 mass% Ni, 99.99 mass% Al and 99.8 mass% Cr. The same result was obtained by [2002Fis] for 550°C using atom probe field ion microscopy. Data of [1993Cot1] concerning solubility of Cr in βNiAl at low temperature shown in Fig. 16 are in a good agreement with [1999Tia] and [2002Fis]. [1997Sch] reported the solubility of Al in γ in $\gamma+\gamma'$ equilibria at 600°C to be less than after [1952Tay2] for 2 to 3 at.% within the whole γ phase boundary extension. The results were obtained by both three-dimensional atom probe and TEM technique on aged low supersaturated alloy 5.2 Al 14.7 Cr (at.%). Isothermal sections at 750, 1000, 1227, 1327, 1427, 1527, 1627, and 1727°C earlier have been calculated by [1974Kau] with some shortcomings in so far as the bcc-phase was calculated to be stable for the entire solution range and no quasibinary behavior along the NiAl–Cr section was revealed. These features have mainly been overcome in the later calculations of the isothermal sections at 1027, 1127, 1227, 1327, 1427 by [1980Cha] and [1985Bar].

Temperature – Composition Sections

Early thermal and microscopic investigations along the vertical Cr–NiAl section by [1953Kor], complemented by hardness and resistivity measurements, revealed a binary eutectic at 1445°C and 33.6Cr–33.2Ni–33.2Al (at.%). This shows a rather restricted solid solubility of the two α and β bcc type phases. [1952Tay2] determined the eutectic $\alpha+\beta$ at 30Ni–30Al (at.%), whereas the findings of [1958Bag1] and [1958Bag2] resulted in 37.4 to 28.2 at.% Cr. More recent studies by [1978Vol] located the binary eutectic in the vertical section Cr–NiAl at 32.5Ni–32.5Al (at.%) and further studies by [1970Cli, 1970Wal,

1971Cli, 1973Wal] and [1984She] arrived at 34Cr-33Ni-33Al (at.%). The microstructure of melt-spun ribbons of the NiAl–Cr eutectic composition (34 at.% Cr) has been studied by optical and TEM microscopy [1984She]. At 1200°C the α Cr phase contains 20.6 mole% NiAl, whereas for the β ,NiAl phase a Cr content of about 8 at.% was estimated by [1984She, 1999Tia]. X-ray diffraction patterns of the CsCl and bcc lattices were said to be indistinguishable. The EMPA data on the mutual solid solubilities within Cr–NiAl section reveal consistently a much larger solid solubility at 1025, 1150 and 1200°C, than earlier reported by [1953Kor, 1958Bag1] and [1958Bag2]. For 1025°C these data are reported by [1982Tu, 1983Lan, 1984Car, 1983Ofo]; for 1150°C - by [1985Ofo]; for 1200°C - by [1984She]. The authors [1953Kor, 1958Bag1] and [1958Bag2] considered the Cr–NiAl section as a quasibinary system, and [1991Rog] presented it in such a way essentially based on [1953Kor, 1958Bag1, 1958Bag2] and [1982Tu, 1984Car, 1983Ofo, 1985Ofo, 1987Ofo]. But the solidus temperature in the α + β phase field is not available and there is no evidence that 1445°C is a maximum solidus temperature of this phase field, which is a necessary condition for the quasibinary section. There are wide ranges of (α Cr) and NiAl solid solutions based on congruent phases coexisting in equilibria in this ternary system. In such a kind of systems, deviation of the quasibinary section from the connected line is a common feature. Therefore it is not surprising that thermodynamic calculations of both [1974Kau] and [2001Dup] revealed no quasibinary behavior along the Ni–AlCr section. This result seems to be confirmed by the experimental points for the set of tie-lines given by [1982Tu, 1983Lan] for 1025°C.

Polythermal Cr–NiAl section according to [2001Dup] is given in Fig. 18. The calculated liquidus and solidus of β phase is in a complete agreement with data [1993Cot1] and α -solvus is in agreement with [1958Bag, 1953Kor]. The Ni₃Al–Ni₃Cr section has been investigated by [1952Tay2, 1953Kor] and [1961Ham]. The experimental isopleth after [1952Tay] is reproduced in Fig. 19a from [1961Ham], who confirmed data [1952Tay2], and calculated by [2001Dup] one is given in Fig. 19b. The isopleth sections 53Cr47Ni - 40Cr45Ni15Al and 49Cr43Ni18Al - 34Cr33Ni33Al, which are close to the monovariant L curves of the L \rightleftharpoons (α Cr)+ β and L \rightleftharpoons (α Cr)+ γ Ni eutectics, are given in Figs. 20, 21 according to [1992Gor] data after minor correction in liquidus-solidus ranges taking into account the ternary system liquidus constitution.

Thermodynamics

The thermodynamic activities of Al in the ternary system for the Cr_{0.0838}Ni_{0.9162}–Al section with aluminium content from 0 to 21 at.% have been determined by [1966Mal] and [1968Mal] using the emf method. The measurements were conducted at temperatures of 1045, 1090, 1135, and 1180 K. The obtained values of excess integral Gibbs energies and entropies of formation of ternary alloys from solid components, as well as activity coefficients at 1045 and 1180 K are presented in Table 3. Because the values of $\Delta_f S^{\text{ex}}$ were determined using the $\Delta_f S^{\text{ex}} = (\Delta G_{T_1}^{\text{ex}} - \Delta G_{T_2}^{\text{ex}})/(T_2 - T_1)$ formula, and T_1 and T_2 are minimum and maximum values (772 and 907°C, respectively), the obtained values of $\Delta_f S^{\text{ex}}$ can be considered rough. The excess integral Gibbs energies of the γ' phase containing 21 at.% Al are -24.27 kJ·mol⁻¹ and -24.77 kJ·mol⁻¹ at 772 and 907°C, respectively. The thermodynamic activities of Al, Cr, Ni at 1423 K in the binaries as well as in the Al–Cr–Ni ternary have been determined by [1983Ofo] and [1985Ofo] using Knudsen cell mass spectroscopy. Some difficulty was experienced with alloys exhibiting high Al vapor pressures, which caused rapid deterioration of the thermocouples and the tantalum used in the construction of the Knudsen cell furnace. The obtained data are presented in Table 4. With applying a defect model, [1988Hoc] calculated the activity of Cr in (NiAl) assuming that NiAl is a compound or a solid solution. The calculated results do not show significant discrepancy. [1990Kek] measured the enthalpy of formation of ternary alloys Cr-72Ni-23Al, Cr-70Ni-22Al and Cr-67Ni-21Al (at.%) at room temperature by solution calorimetry. It was reported that the alloys were of γ' single-phase, according to X-ray and metallographic examination results. The temperature of heat treatment, however, was not given. The enthalpies of formation of these three alloys at 298 K are 36.7 ± 1.4 kJ·mol⁻¹, 35.2 ± 1.4 kJ·mol⁻¹ and 36.7 ± 1.4 kJ·mol⁻¹, respectively. [1992Hil] studied the vaporization of the Ti-1.85Ta-3.9W-2.05Mo-14.9Cr-70.3Ni-4.5Al (mass%) alloy using Knudsen effusion mass spectrometry in the temperature range 1393–1562 K. The partial pressures of Cr, Ni and Al were determined in this temperature range (Table 5). Chemical activities and excess chemical potentials

were evaluated from these pressures for a temperature of 1500 K (Table 6). [2000Sal] determined the partial and the integral enthalpies of mixing of liquid constituent ternary alloys along four isopleths $x_{\text{Cr}}/x_{\text{Ni}}$, namely, 0.18, 0.43, 1.00 and 4.00, by high temperature isoperibolic calorimetry at $1454 \pm 5^\circ\text{C}$. The results were analytically described by the thermodynamically adapted power series according to the Kohler interpolation geometry and to a regular association model (RAM). The maximal discrepancies between experimental and calculated values do not exceed $1 \text{ kJ}\cdot\text{mol}^{-1}$, which is within the confidence bands of the data measured. On the basis of data on phase equilibria and thermodynamic properties, a new thermodynamic description was developed by [1999Hua] by thermodynamic modeling. The liquid, bcc, and fcc phases were modeled as substitutional solutions. No ternary phases were included. The comparison between the calculation and experimental data are presented. The calculated and experimentally determined activity data of Cr, Ni and Al at 1150°C [1983Ofo, 1985Ofo] as well as the enthalpies of formation of series of the alloys at 25°C [1990Kek] are in a good agreement. Thermodynamic modeling of the ternary system was done by [2002Bro] using Thermo-Calc software package. Two different sets of thermodynamic parameters of Al–Cr–Ni system for two types of thermodynamic models were published for the first time in the literature. The comparison shows acceptable agreement between the calculated and experimental data.

Notes on Materials Properties and Applications

[1960Kor] showed an increase in hardness values for Al–10 mass% Cr–Ni alloys from $1300 \text{ MN}\cdot\text{m}^{-2}$ at 3 mass% Al to $2500 \text{ MN}\cdot\text{m}^{-2}$ at 8 mass% Al. Hardness-temperature curves have been measured by [1959Gua] on Ni_3Al –2.5 and 7.5 at.% Cr alloys. [1999Tia] studied the hardness variations during aging of β phases based alloys using a micro-Vickers hardness tester. It was established that typical age-hardening and over-age softening occur by fine precipitation of αCr in NiAl under aging after quenching from high temperatures. The degree of age hardening of stoichiometric NiAl in which equal amounts of Ni and Al are replaced by Cr is smaller than that obtained in off-stoichiometric NiAl. [1983Dan] investigated the influence of the microstructure in the cast 70 AlCrNi alloy on its ductility during deformation and [2001Com] investigated the effect of alloys microstructure on Vicker's microhardness in the aluminium-rich corner. Compression creep behavior was investigated in ternary γ' single crystals containing Cr with stress axes parallel to the crystallographic orientation near [001] [1991Fie]. Some mechanical properties of the intermetallics based alloys were investigated by [1991Miu, 1993Cot2, 2002Guo]. [1987Nes1, 1987Nes2] and [1987Nes3] studied interdiffusion in Ni-rich Al–Cr–Ni alloys at 1100 and 1200°C . Concentration dependent ternary interdiffusion coefficients were derived from EMPA studies of γ/γ' and $\gamma/\gamma+\beta$ diffusion couples, and a ternary finite-difference interdiffusion model was employed to predict concentration profiles, which at 1100 and 1200°C were in a good agreement with the measurements. [1997Kai] investigated pseudo-interface formation as well as diffusion behavior in the B2 phase region of NiAl-based diffusion couples (concentration ranges 30–60 at.% Al, 0–10 at.% Cr) using EMPA. Also the isoactivity lines for alloying elements in the B2 phase were calculated. In the CALPHAD summary [1999Kau], results of studying diffusion couples at 1200°C combined with calculations based on thermo-calc and DICTRA Programs are presented. Assessments of diffusional mobilities were carried for face-centered cubic alloys [1996Eng] and for Ni-base superalloys [2002Cam]. Models for simulating diffusion in multiphase dispersion were applied by [1997Eng] and [1998Mat] to multiphase diffusion couples at 1200°C . A general equation for the effective diffusivity (D^{eff}) in a two-phase $\gamma+\beta$ region was derived by [1999Che] based on the assumption that the second phase can be treated as a point source or sink of solute. The variation of D^{eff} in the $\gamma+\beta$ region was calculated and shown to be up to 70 %. [1978Vol] and [1986Gor] examined the possibility of obtaining a directionally crystallized eutectic, whereas the effect of alloy additions on the rod-plate transition in the directionally solidified NiAl–Cr eutectic was studied by [1970Cli] and dislocation networks at the interface between phases were studied by [1971Cli]. The rod-plate transition was observed to occur near the composition for zero lattice mismatch. The average coefficient of thermal expansion of NiAl was $1.4\cdot 10^{-5}$ and that of Cr was $1.1\cdot 10^{-5}$ as obtained from X-ray diffraction in the temperature range from room temperature to 1000°C [1970Cli]. Similar investigations concern the effect of solidification rate on structure and high temperature strength [1970Wal], on the stability of the directionally solidified Cr–NiAl eutectic [1973Wal] as well as the effect

of rapid solidification rate on the microstructure and phase solubility in Cr–NiAl eutectic [1985Boe]. A two-dimensional model for solidification and secondary phase precipitation in directionally solidified superalloys is presented by [2000Boe], with application to Al–Cr–Ni as a ternary model system. [1960Kor] established that the electric resistivity for Al–10 mass% Cr–Ni alloys increases from $0.85 \cdot 10^{-4} \Omega\text{m}$ at 3 mass% Al to $1.5 \cdot 10^{-4} \Omega\text{m}$ at 8.5 mass% Al. [1972Sch] studied the polarization relations to quantitatively isolate the γ' phase by electrochemical treatment from Al–Cr–Ni alloys with more than 70 at.% Al. The isolated particles were examined by X-ray and electron microscopy. The lattice parameters did not change with changing the form or size of the particles. Dependence of the magnetic susceptibility on the temperature in the interval 0–1000°C of the alloy $\text{Cr}_{6.5}\text{Ni}_{46.1}\text{Al}_{47.4}$ (at.%) was established by [1999Zya]. No long-range ordering is observed in the Ni_3Al –Cr Ni_3 phase in the temperature range 893–1073°C [1961Ham]. In high-Al-content alloys, the γ' precipitation rate is extremely high and cannot be suppressed by ice-brine quenching; the creep strength and yield strength of Cr Ni_3 – Ni_3Al alloys at 750°C increased five or six fold as a result of γ' precipitation [1961Ham]. Atom probe field-ion microscopy (APFIM) data were used by [1984Hil] to differentiate the spinodal decomposition mode in a Ni–14Al (at.%) alloy and a nucleation, growth and coarsening mechanism for the two-phase microstructure of a Ni–20Cr–14Al (at.%) alloy in the quenched state at 620°C [1984Hil]. High temperature X-ray diffraction was used by [1971Cli] (25 to 1000°C) and [1976Low] (25 to 1200°C) to study the temperature variation in the unit cell dimensions for the α , β , γ , and γ' phases in a series of Cr–Ni–Al alloys ranging from 6 to 30 at.% Al and 10 to 22 at.% Cr [1976Low], and for β ,NiAl in a directionally solidified eutectic alloy containing 34 vol% Cr rods of 2 μm diameter in the NiAl matrix [1971Cli]. Unit cell dimensions were found to follow an empirical thermal expansion equation $a(T) = a(250^\circ\text{C})(1+R)(1+T/293)^{1.5}$ where the expansion constant R was given as $13.4 \cdot 10^{-4}$ for αCr , $19.2 \cdot 10^{-4}$ for γ/γ' and $R = 19.9 \cdot 10^{-4}$ for βNiAl [1976Low]. [1996Sch] and [1997Sch] presented results of atomic scale investigation of ordering and precipitation of the γ' phase in a model Al–Cr–Ni alloy, composition of which was 15Cr80Ni5Al (at.%) or 14.7Cr80.1Ni5.2Al (at.%), respectively, aged at 600°C. Both three-dimensional (3D) atom probe (AP) and TEM techniques were used. It was shown in both publications that small γ' particles, 2 nm in diameter, form already after 1 h. Even for such very small sizes, particles are well ordered. Their composition is close to the equilibrium composition of the γ' phase and does not evolve as ageing proceeds. The time evolution of the particle size and of the composition of the γ matrix suggests that a growth mechanism is involved before 4 h, followed by a coarsening regime. The alloys on the base of the γ' phase were also studied by [1997Kre], [1997Sai] and [2001Ter]. [1997Kre] established the presence of the hypostoichiometric γ' Ni_3Al phase in chill-cast $\text{Ni}_{20}\text{Cr}_{13}\text{Al}$ alloy using XRD and TEM. The spheroid particles of γ' change their shape to cubical at the very beginning of annealing at 800°C. This change is accompanied by a decrease in volume fraction of γ' and at the same time by an increase in hardness. Such annealing also leads to the precipitation of Cr-rich α phase, starting at grain boundaries. The Monte-Carlo method was applied by [1997Sai] to the simulation of temporal evolutions of atomic arrangement in Ni-base ternary alloy Cr3.0Ni73.5Al23.5 (at.%). The ordering of the fcc phase to L_{12} structure is controlled by randomly selecting a single atom with one of its neighboring atoms. [2001Ter] proposed a method of the site preference determination of substitutional elements in intermetallic compounds through the measurements of thermal conductivities. The authors note that the laser-flash method for thermal conductivity measurement is very convenient for intermetallic compounds, since the preparation of a coin shaped specimen for the laser-flash measurement is very easy to achieve even for brittle materials. It is demonstrated in Ni_3Al –2Cr (at.%) alloy that the ridge direction in thermal conductivity contours in the ternary γ' phase agrees with that of the solubility lobe of the γ' phase in ternary phase diagrams. [1999Tia, 1999Zak] and [2002Fis] investigated β phase based alloys. In particular, [1999Tia] and [2002Fis] studied the precipitation of (αCr) in B2-ordered β ,NiAl using TEM method [1999Tia] and atom probe field ion microscopy [2002Fis]. According to [1999Tia], by aging at temperatures around 1073 K after solution annealing at 1290°C, fine spherical particles of αCr appear in the NiAl matrix between the (αCr), which rapidly precipitate during quenching. Atom location by channeling enhanced microanalysis technique has been used to determine the site occupancy of Cr in NiAl. [2002Fis] studied the alloys (at.%): $\text{Cr}_2\text{Ni}_{48}\text{Al}_{48}$, $\text{Cr}_2\text{Ni}_{49}\text{Al}_{49}$ and $\text{Cr}_2\text{Ni}_{48}\text{Al}_{50}$, annealed at 550°C for 100 h, and precipitates of (αCr) were detected in the first two alloys ranging from few nanometers in size. As to $\text{Cr}_2\text{Ni}_{48}\text{Al}_{50}$ samples, different precipitates

containing 77 at.% Cr and 23 at.% Al are observed in it. [1999Zak] established temperature coefficients of linear and volume expansion of the $\text{Cr}_{0.14}\text{NiAl}_{1.03}$ alloy at temperature ranges of 25 to 760°C and 760 to 1000°C using high-temperature XRD studies. [1999Zya] investigated the electronic structure of the $\text{Cr}_{6.5}\text{Ni}_{46.1}\text{Al}_{47.4}$ (at.%) alloy. Mechanism of hot corrosion of the Al–Cr–Ni alloys were investigated by [1993Gle] and [1996Lon]. Development of oxidation-resistant high-temperature intermetallics and corrosion-resistant coatings was carried out by [1992Gra] and [1993Nic]. In particular, it was established that an increase in the Cr content of two-phase NiAl–Cr alloys reduces the oxidation resistance [1992Gra].

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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
(Al) ≤ 660.452 $\text{Cr}_x\text{Ni}_y\text{Al}_{1-x-y}$	$cF4$ $Fm\bar{3}m$ Cu	$a = 404.88$	pure Al, $T = 24^\circ\text{C}$ [V-C] $x = 0, y = 0$ to 0.004 [2003Sal] $y = 0, x = 0$ to 0.00375 [2003Cor]
α , (Cr) ≤ 1863 α , $\text{Cr}_{1-x-y}\text{Ni}_y\text{Al}_x$	$cI2$ $Im\bar{3}m$ W	$a = 288.4$	pure Cr, $T = 27^\circ\text{C}$ [V-C] $x = 0, y = 0$ to 0.31 [Mas2] $y = 0, x = 0$ to 0.46 [1980Sch]
γ , (Ni) < 1455 $\text{Cr}_y\text{Ni}_{1-x-y}\text{Al}_x$	$cF4$ $Fm\bar{3}m$ Cu	$a = 352.40$ $a = 353.4$ $a = 353.8$ $a = 354.5$ $a = 355.1$ $a = 356.0$ $a = 355.8$ $a = 356.5$	pure Ni, $T = 25^\circ\text{C}$ [1984Och2, Mas2] $x = 0, y = 0.1$ [1952Tay2] $x = 0, y = 0.15$ [1952Tay2] $x = 0, y = 0.2$ [1952Tay2] $x = 0, y = 0.25$ [1952Tay2] $x = 0, y = 0.25$ [1984Och2] $x = 0, y = 0.3$ [1984Och2] $x = 0, y = 0.35$ [1984Och2]
		$a = 352.8$ $a = 353.8$ $a = 355.1$ $a = 355.8$ $a = 356.3$ $a = 353.2$ $a = 353.7$ $a = 354.1$ $a = 354.8$ $a = 355.4$ $a = 355.9$ $a = 356.5$	Slowly cooled alloys, $x = 0.025$ or $0.05, y = 0$ to 0.3 [1952Tay2]: $x = 0.025, y = 0$ $x = 0.025, y = 0.12$ $x = 0.025, y = 0.22$ $x = 0.025, y = 0.27$ $x = 0.025, y = 0.32$ $x = 0.05, y = 0$ $x = 0.05, y = 0.05$ $x = 0.05, y = 0.1$ $x = 0.05, y = 0.15$ $x = 0.05, y = 0.2$ $x = 0.05, y = 0.25$ $x = 0.05, y = 0.3$
			[1985Ofo]: $x = 0$ to $0.16, y = 0$ to $0.48, T = 1150^\circ\text{C}$ [1982Tu]: $x = 0$ to $0.14, y = 0$ to $0.43, T = 1025^\circ\text{C}$ [1952Tay2]: $x = 0$ to $0.4, y = 0$ to $0.4, T = 850^\circ\text{C}$ $x = 0$ to $0.38, y = 0$ to $0.37, T = 750^\circ\text{C}$
		$a = 356.4$	Quenched from 1200°C [1961Pry]: $x = 0.17, y = 0.111$
		$a = 354.6$ $a = 355.3$ $a = 353.5$ $a = 362.3$	Quenched from $T = 1000^\circ\text{C}$: $x = 0, y = 0.2$ [1955Kor1] $x = 0.1, y = 0.2$ [1955Kor1] $x = 0, y = 0.111$ [1961Pry] $x = 0.14, y = 0.111$ [1961Pry]
	Sublattice of γ , (Ni), order-disorder transformation at $\sim 500^\circ\text{C}$	-	$x = 0.05, y = 0.4$ [1952Tay1], adapted from [1991Rog]

Phase/ Temperature Range [°C]	Pearson Symbol/ Space Group/ Prototype	Lattice Parameters [pm]	Comments/References
θ , $\text{Cr}_2\text{Al}_{13}$ ≤ 791	<i>mC</i> 104 <i>C2/m</i> V_7Al_{45}	$a = 2519.6$ $b = 757.4$ $c = 1094.9$ $\beta = 128.7^\circ$	Room temperature, 13.5 at.% Cr [1975Ohn, 1995Aud, 2003Cor] Sometimes called CrAl_7 in the literature
θ , $(\text{Cr,Ni,Al})_2(\text{Al,Cr,Ni})_{13}$	Monoclinic <i>C2/m</i> (?) Triclinic <i>P\bar{3}m</i> (?)	$a = 2525.6$ $b = 758.2$ $c = 1095.5$ $\beta = 128.68^\circ$ $a = 1100$ $b = 2470$ $c = 1320$ $\beta = 113.85^\circ$ $a = 1100$ $b = 1905$ $c = 1320$ $\alpha = 79.2^\circ$ $\beta = 113.85^\circ$ $\gamma = 78.3^\circ$	From 0 to ~2.5 at.% Ni, as-cast [2001Com] In the alloys (at.%) $\text{Cr}_4\text{Ni}_4\text{Al}_{92}$ and $\text{Cr}_{10}\text{Ni}_3\text{Al}_{87}$ quenched from liquidus temperature [1996Ros] In the alloy (at.%) $\text{Cr}_4\text{Ni}_4\text{Al}_{92}$ quenched from liquidus temperature, two polytypes, resulting in a periodic (111) twinning of θ [1996Ros]
ρ , $\text{Cr}_2\text{Al}_{11}$ ≤ 941	<i>oC</i> 584 <i>Cmcm</i>	$a = 1240$ $b = 3460$ $c = 2020$ $a = 1260$ $b = 3460$ $c = 2000$	16.9 to 19.2 at.% Cr [1995Aud, 2000Mah, 2003Cor] Sometimes called CrAl_5 in the literature Quenched from 920°C, 16.9 to 19.2 at.% Cr [1995Aud, 2000Mah, 2003Cor] “ ϵCrAl_4 ” [1992Wen]
ρ , $(\text{Cr,Ni,Al})_2(\text{Al,Cr,Ni})_{11}$			From 0 to ~5.5 at.% Ni, as-cast [2001Com]
μ , CrAl_4 ≤ 1031 μ , $\text{Cr}_{1-x}\text{Al}_x$	<i>hP</i> 574 <i>P6_3/mmc</i> μMnAl_4	$a = 1998$ $c = 2467$ $a = 2010$ $c = 2480$	$x = 0.78$ to 0.80 [2003Cor] $x = 0.788$ to 0.794, $T = 800^\circ\text{C}$ [1995Aud] $x = 0.791 \pm 0.003$ [1995Aud, 2000Mah, 2003Cor] $x = 0.777 \pm 0.001$, quenched from 1000°C [2000Mah, 2003Cor]
μ , $(\text{Cr,Ni,Al})_1(\text{Al,Cr,Ni})_4$			From 0 to ~4 at.% Ni, as-cast [2001Com]
i , CrAl_4 (or CrAl_5)	icosahedral	-	In spinning alloys Al–Cr at 8 to 13 at.% Cr; or in transformed amorphous of 20 at.% Cr, metastable [1998Mur]
ϵ_1 , Cr_4Al_9 (h) ϵ_1 , $\text{Cr}_{1-x}\text{Al}_x$ ~1170 to ~1060 ϵ_1 , $(\text{Cr,Ni,Al})_4(\text{Al,Cr,Ni})_9$	<i>cI</i> 52 <i>I\bar{4}3m</i> Cu_4Al_9	$a = 912.3$	$x \approx 0.65$ to 0.70 [Mas2] $x = 0.71$ at Al-rich limit, quenched from 920°C [1995Aud] From 0 to ~3.3 at.% Ni, as-cast [2001Com] From 0 to 3.5 at.% Ni; 3.5 at.% Ni at 60 at.% Al, annealed at 1025°C [1982Tu]
ϵ_2 , Cr_4Al_9 (r) $\text{Cr}_{1-x}\text{Al}_x$ ≤ 1060	<i>hR</i> 52 <i>R\bar{3}m</i> Cr_4Al_9	$a = 1291$ $c = 1567.7$	$x = 0.650$ to 0.69 [Mas2] [1968Lin, Mas2]

Phase/ Temperature Range [°C]	Pearson Symbol/ Space Group/ Prototype	Lattice Parameters [pm]	Comments/References
ϵ_3 , Cr_4Al_9 (r) $\text{Cr}_{1-x}\text{Al}_x$ ≤ 700	-	-	$x \approx 0.67$ to 0.69 [Mas2]
ζ_1 , Cr_5Al_8 (h) ζ_1 , $\text{Cr}_{1-x}\text{Al}_x$ ~ 1350 to ~ 1100 ζ_1 , (Cr,Ni,Al) ₅ (Al,Cr,Ni) ₈	<i>cI52</i> <i>I43m</i> Cu_5Zn_8	$a = 904.7$ to 910.4	$x = 0.58$ to 0.65 quenched from liquid [1989Ell] From 0 to ~ 3.3 at.% Ni; 3.3 at.% Ni at ~ 60 at.% Al, as-cast [2001Com]
ζ_2 , Cr_5Al_8 (r) $\leq 1100(?)$ ζ_2 , $\text{Cr}_{1-x}\text{Al}_x$	<i>hR26</i> <i>R3m</i>	$a = 1271.9$ $c = 793.6$ $a = 1276.5$ to 1271.5 $c = 795.4$ to 782.8 $a = 1272.8$ $c = 794.2$	[1994ICD], No. 29-15 $x = 0.58$ to 0.65 [1989Ell] [1977Bra]
η , Cr_2Al < 910	<i>tI6</i> <i>I4/mmm</i> MoSi_2	$a = 300.45$ $c = 864.77$ $a = 300.5(1)$ $c = 864.9(1)$	~ 65.5 to 71.4 at.% Cr [1998Mur] [1937Bra, 1998Mur] [1989Ell]
X ≤ 400	Cr_5Al_3 or Cr_3Al superlattice	-	Possibly metastable [1998Mur] ~ 75 to ~ 80 at.% Cr [1981Bro, 1981Ten]
ω	-	-	In quenched alloys Cr-Al at 60 to 100 at.% Cr, like metastable ωTi in Ti alloys [2000Sha1, 2000Sha2]
π , NiAl_3 < 856	<i>oP16</i> <i>Pnma</i> NiAl_3	$a = 661.15$ $b = 736.64$ $c = 481.18$ $a = 661.3 \pm 0.1$ $b = 736.7 \pm 0.1$ $c = 481.1 \pm 0.1$	[L-B] [1996Vik] [1997Bou, V-C]
π , (Ni,Cr)Al ₃	<i>oP16</i> <i>Pnma</i> CFe_3 Modulated	$a = 659.8$ $b = 735.1$ $c = 480.2$ $a = 660$ $b = 740$ $c = 480$	 [1997Bou, V-C] In the alloys (at.%) $\text{Cr}_9\text{Ni}_9\text{Al}_{82}$ and $\text{Cr}_4\text{Ni}_4\text{Al}_{92}$, as-cast [1996Ros]
δ , Ni_2Al_3 < 1138		$a = 403.63$ $c = 490.65$ $a = 402.8$ $c = 489.1$	36.8 to 40.5 at.% Ni [Mas2] [L-B] [1997Bou, V-C]

Phase/ Temperature Range [°C]	Pearson Symbol/ Space Group/ Prototype	Lattice Parameters [pm]	Comments/References
δ , (Ni,Cr,Al) ₂ (Al,Cr,Ni) ₃			From 0 to ~17 at.% Cr, as-cast [2001Com] From 0 to 3 at.% Cr; 3 at.% Cr at 36 at.% Ni, annealed at 1025°C [1982Tu] In the alloy Cr _{0.14} NiAl _{1.03} [1999Zak]: $T = 25^\circ\text{C}$ $a = 404.5$ $c = 489.5$ $a = 406$ $T = 400^\circ\text{C}$ $c = 492$ $a = 408$ $T = 850^\circ\text{C}$ $c = 494$ $a = 409$ $T = 1000^\circ\text{C}$ $c = 496$ $a = 410$ In the alloy Cr _{6.5} Ni _{46.1} Al _{47.4} , $T = 1000^\circ\text{C}$ $c = 493$ [1999Zak]
Ni ₃ Al ₄ < 702	$cI112$ $Ia\bar{3}d$ Ni ₃ Ga ₄	$a = 1140.8(1)$	[1989Ell, V-C]
β , NiAl < 1676	$cP2$ $Pm\bar{3}m$ CsCl	$a = 287.04$ $a = 287.26$ $a = 286.0$ $a = 287.0$ $a = 288.72(2)$ $a = 287.98(2)$ $a = 289.0$ $a = 289.7$ $a = 290.4$ $a = 291.2$ $a = 291.9$ $a = 293.2$ $a = 290.2$ $a = 290.9$	42 to 69.2 at.% Ni [Mas2] 57.7 at.% Ni [L-B] 46.6 at.% Ni [L-B] [1987Kha] 63 at.% Ni [1993Kha] 50 at.% Ni [1996Pau] 54 at.% Ni [1996Pau] [1971Cli]: $T = 0^\circ\text{C}$ $T = 200^\circ\text{C}$ $T = 400^\circ\text{C}$ $T = 600^\circ\text{C}$ $T = 800^\circ\text{C}$ $T = 1000^\circ\text{C}$ 33 at.% Cr, $T = 600^\circ\text{C}$ [1971Cli] 33 at.% Cr, $T = 800^\circ\text{C}$ [1971Cli] 0 to ~8 at.% Cr; ~8 at.% Cr at ~50 at.% Ni, annealed at 1290°C [1999Tia] 0 to ~12 at.% Cr; ~12 at.% Cr at ~57 at.% Ni, $T = 1200^\circ\text{C}$ [1984Car] ~8 at.% Cr at ~62 at.% Ni, $T = 1127^\circ\text{C}$ [1989Hon] 8 at.% Cr at 58 at.% Ni, annealed at 1025°C [1982Tu] 0 to < 2 at.% Cr at 48 to 50 at.% Ni, annealed at 550°C [2002Fis]
Ni ₅ Al ₃ < 723	$oC16$ $Cmmm$ Pt ₅ Ga ₃	$a = 753$ $b = 661$ $c = 376$	63 to 68 at.% Ni [1993Kha, Mas2] at 63 at.% Ni [1993Kha]

Phase/ Temperature Range [°C]	Pearson Symbol/ Space Group/ Prototype	Lattice Parameters [pm]	Comments/References
γ' , Ni_3Al <1372 γ' , $(\text{Ni},\text{Cr},\text{Al})_3(\text{Al},\text{Cr},\text{Ni})_1$	$cP4$		73 to 76 at.% Ni [Mas2]
	$Pm\bar{3}m$	$a = 356.6$	[1952Tay2]
	Cu_3Au	$a = 357.0$	[1984Och2, 1959Gua]
		$a = 356.77$	[1986Hua]
		$a = 356.32$	disordered [1998Rav]
		$a = 357.92$	ordered [1998Rav]
			From 0 to 5 at.%Cr; 5 at.%Cr at 72 at.%Ni, 1300°C [1994Jia, 2001Dup]
			From 0 to 7.5 at.%Cr; 7.5 at.%Cr at 70 at.% Ni, 1200°C [1984Car, 1998Qia, 2001Dup]
			From 0 to 10 at.% Cr; 10 at.% Cr at 70 at.% Ni, 1127°C [1989Hon]
			From 0 to 16 at.% Cr; 16 at.% Cr at 65 at.% Ni, annealed at $T = 1025^\circ\text{C}$ [1982Tu]
		$a = 356.3$	2.5 at.% Cr [1952Tay2]
		$a = 356.5$	5 at.% Cr [1952Tay2]
Ni_2Al_9	$mP22$ $P2_1/c$ Ni_2Al_9	$a = 868.5(6)$	Metastable
		$b = 623.2(4)$	[1988Li, 1997Poh]
		$c = 618.5(4)$	
		$\beta = 96.50(5)^\circ$	
$\text{Ni}_x\text{Al}_{1-x}$ $0.60 < x < 0.68$	$tP4$ $P4/mmm$ AuCu	$a = 383.0$	Martensite, metastable
		$c = 320.5$	[1993Kha]
		$a = 379.5$	62.5 at.% Ni [1991Kim]
		$c = 325.6$	
		$a = 379.5$	63.5 at.% Ni [1991Kim]
		$c = 325.6$	
		$a = 379.5$	66.0 at.% Ni [1991Kim]
		$c = 325.6$	
		$a = 379.9$ to 380.4	64 at.% Ni [1997Pot]
		$c = 322.6$ to 323.3	
	m^{**}	$a = 371.7$ to 376.8	65 at.% Ni [1997Pot]
		$c = 335.3$ to 339.9	
		$a = 378.00$	[1998Sim]
		$c = 328.00$	
		$a = 418$	[1992Mur]
		$b = 271$	
		$c = 1448$	
		$\alpha = 90^\circ$	
		$\beta = 93.4^\circ$	
		$\gamma = 90^\circ$	

Phase/ Temperature Range [°C]	Pearson Symbol/ Space Group/ Prototype	Lattice Parameters [pm]	Comments/References
Ni ₂ Al	<i>hP3</i> <i>P3m1</i> CdI ₂	$a = 407$ $b = 499$	Metastable [1993Kha]
	a^*126 <i>P1</i>	$a = 1252$ $b = 802$ $c = 1526$ $\alpha = 90^\circ$ $\beta = 109.7^\circ$ $\gamma = 90^\circ$	[1994Mur]
D ₁		-	Metastable [1988Li]
D ₄		-	Metastable [1988Li]
CrNi ₂	<i>oP6</i> <i>Immm</i> MoPt ₂ (?)	$a = 252.4$ $b = 757.1$ $c = 356.8$	60 to 76.5 at.% Ni [Mas2] [P]
*Cr ₁₅ Ni ₁₀ Al ₇₅ ≤ 940	-	-	[1982Tu], adapted from [1984Mer1, 1984Mer2]
*τ ₁	Monoclinic, <i>P2₁</i> or <i>P2₁/m</i>	$a = 1340$ $b = 1255$ $c = 1255$ $\beta = 100^\circ$	Called “φ” 15.7 to 17.9 at.% Cr, 5.6 to 9.7 at.% Ni, 73.3 to 78.5 at.% Al [1996Ros] Cr _{17.9} Ni _{5.6} Al _{76.5} , quenched from liquid, [1996Ros]
*τ ₁ '	Orthorhombic polytype, <i>Pn2₁a</i>	$a = 1255$ $b = 1255$ $c = 2640$	Cr _{17.9} Ni _{5.6} Al _{76.5} , quenched from liquid, results in a periodic twinning of τ ₁ [1996Ros]
*τ ₁ , Cr _x Ni _y Al _z			Cr ₁₆ Ni ₈ Al ₇₆ , annealed at 840°C (5 h) [1990Col]; Cr _{17.0} Ni _{9.7} Al _{73.3} , annealed at 800°C (6 d) [1996Ros] Cr _{17.0} Ni _{9.7} Al _{73.3} Cr _{15.7} Ni _{5.8} Al _{78.5} , as cast [1996Ros]
*τ ₂			Called “λ” [1996Ros] 11.7 to 12.8 at.% Cr, 8.5 to 10.4 at.% Ni, 76.8 to 79 at.% Al
*τ ₂ '	Orthorhombic <i>Immm</i> (?)	$a = 1255$ $b = 1255$ $c = 3075$	Cr _{12.5} Ni _{8.5} Al _{79.0} , quenched from liquid [1996Ros]
	Triclinic superstructure	$a = 1255$ $b = 2510$ $c = 1775$ $\alpha = \beta = 110.4^\circ$ $\gamma = 90^\circ$	Cr _{12.5} Ni _{8.5} Al _{79.0} , quenched from liquid or from liquidus during DTA experiment [1996Ros]
*τ ₂ , Cr _x Ni _y Al _z			Cr _{12.4} Ni _{9.5} Al _{78.1} and Cr _{11.7} Ni _{9.3} Al ₇₉ , as cast Cr _{12.8} Ni _{10.4} Al _{76.8} , in the alloy Cr ₂ Ni ₂₃ Al ₇₅ annealed at $T = 830^\circ\text{C}$ (7 d) [1996Ros]

Phase/ Temperature Range [°C]	Pearson Symbol/ Space Group/ Prototype	Lattice Parameters [pm]	Comments/References
* τ_3 (h)	Rhombohedral $R\bar{3}$ or $R3$ or hexagonal system	$a = 2870$ $\alpha = 36^\circ$ $a = 1770$ $c = 8040$	Called “ ρ_1 ” $\text{Cr}_{18.7}\text{Ni}_{1.7}\text{Al}_{79.6}$, quenched from liquid [1996Ros]
τ_4 (h)	Hexagonal, $P6_3$	$a = 1770$ $c = 1240$	Called “ ρ_2 ” $\text{Cr}_{18.7}\text{Ni}_{1.7}\text{Al}_{79.6}$, quenched from liquid [1996Ros]
* τ_5 (...)	Hexagonal $P6_3$	$a = 3070$ $c = 1240$ $a = 1767.4$ $c = 1251.6$	Called “ ρ_3 ” $\sim\text{Cr}_{18.7}\text{Ni}_{1.7}\text{Al}_{79.6}$, quenched from liquidus during DTA experiment [1996Ros] $\sim\text{Cr}_{18}\text{Ni}_6\text{Al}_{76}$, called “ κ ”, single crystal separated from slowly cooled $\text{Cr}_{10}\text{Ni}_{10}\text{Al}_{80}$ alloy [1997Li, 1997Sat]; in conventionally solidified alloy $\text{Cr}_{15}\text{Ni}_5\text{Al}_{80}$
* τ_6 , “H-(CrNiAl)”	Hexagonal	$a = 1240$ $c = 1240$	Metastable phase in the rapidly solidified alloy $\text{Cr}_{0.5}\text{Ni}_{0.5}\text{Al}_6$ together with icosahedral and decagonal quasicrystals [1989Zho]
* i			Metastable [1989Zho]
* d			Metastable [1989Zho], analogous to the Al–Mn–Ni phase [1988Ten]

Table 2: Invariant Equilibria

Reaction	T [°C]	Type	Phase	Composition (at.%)		
				Al	Cr	Ni
$L \rightleftharpoons \alpha + \beta$	> 1445	e_1	-	-	-	-
$L + \gamma' \rightleftharpoons \beta + \gamma$	1350	U_1	L	23.6	4.5	71.9
			γ'	23.4	2.9	73.7
			γ	20.4	4.2	75.4
			β	28.4	2.4	69.2
$L \rightleftharpoons \alpha + \beta + \gamma$	1300 ± 20	E_1	L	17.4	32.5	50.1
			α	3.1	79.9	16.9
			β	30.2	14.5	55.2
			γ	12.9	36.2	50.9
$L + \alpha \rightleftharpoons \beta + \zeta$?	U_2	L	?	?	?
			α	~34.0	~63.0	~3.0
			β	~51.0	~9.0	~40.0
			ζ	~41.0	~52.0	~7.0
$L + \beta \rightleftharpoons \delta + \zeta$?	U_3	L	?	?	?
			β	~40.0	~5.0	~55.0
			δ	~37.0	~3.0	~60.0
			ζ	61.9	36.3	~1.8

Reaction	$T [^{\circ}\text{C}]$	Type	Phase	Composition (at.%)		
				Al	Cr	Ni
$\gamma + \beta \rightleftharpoons \gamma' + \alpha$	996	U_4	γ	11.7	28.1	60.2
			β	30.5	9.6	59.8
			γ'	21.2	10.8	68.0
			α	0.2	96.3	3.5
$L \rightleftharpoons \theta + (\text{Al}) + \pi$	634	E_2	-	-	-	-

Table 3: Integral Excess Gibbs Energies, Entropies of Formation of Al–Cr–Ni Alloys and Activity Coefficients of Aluminium at 1045 and 1180 K [1966Mal, 1968Mal]

x_{Al}	$\Delta_f G^{\text{ex}} [\text{J}\cdot\text{mol}^{-1}]$		$\Delta_f S^{\text{ex}} [\text{J}\cdot(\text{K}\cdot\text{mol})^{-1}]$	$\lg \gamma_{\text{Al}}$	
	1045 K	1180 K		1045 K	1180 K
0	–670	–590	–0.84	–5.53	–4.65
0.01	–1755	–1630	–1.26	–4.72	–3.86
0.05	–5560	–5400	–2.51	–4.39	–3.69
0.07	–7400	–7060	–2.51	–3.75	–3.37
0.11	–10630	–10250	–2.51	–3.68	–3.29
0.13	–12120	–11790		–3.69	–3.23
0.15	–13600	–13260			
0.21	–17870	–17530			

Table 4: Activity Data of Cr, Ni and Al for a Temperature 1423 K (the Reference States are fcc–Ni, bcc–Cr and Liquid Aluminium) [1983Ofo, 1985Ofo]

Alloy compositions (at.%)			Activities a_i		
Cr	Ni	Al	Cr	Ni	Al
8.4	63.5	28.1	0.441	0.343	0.0003
19.39	20.51	60.1	0.18	<0.0001	0.34
19.9	59.7	20.4	0.644	0.067	0.003
20.3	58.8	20.9	0.511	0.392	0.0006
20.6	37.2	42.2	0.569	0.036	0.014
37.3	19.9	42.8	0.629	0.021	0.231
38.1	39.5	22.4	0.566	0.142	0.0004
38.7	41.9	19.4	0.603	0.219	0.0008
39.7	15.9	44.4	0.3	<0.001	0.29
59.7	20.4	19.9	0.644	0.067	0.003

Table 5: Partial Pressures of Cr, Ni and Al over the Alloy Ti-1.85Ta-3.9W-2.05Mo-14.9Cr-70.3Ni-4.5Al (mass%) and Temperature Range of the Measurements ΔT [1992Hil]

Gaseous species i	ΔT [K]	$\ln(p_i/\text{Pa}) = -A \cdot 10^4/T + B$		P_i [Pa] at 1500 K
		A	B	
Cr	1393-1463	3.941 ± 0.096	22.56 ± 0.71	$2.6 \cdot 10^{-2} \pm 19 \%$
	1463-1562	4.264 ± 0.092	24.77 ± 0.58	
Ni	1393-1463 ^{*)}	4.821 ± 0.215	26.0 ± 1.47	$2.4 \cdot 10^{-3} \pm 21 \%$
	1463-1562	5.219 ± 0.080	28.73 ± 0.43	
Al	1443-1463	3.810 ± 0.879	13.10 ± 6.05	$3.5 \cdot 10^{-6} \pm 49 \%$
	1463-1562	5.322 ± 0.621	22.88 ± 3.79	

^{*)} 1463 K is the solvus temperature

Table 6: Chemical Activities a_i and Excess Chemical Potentials μ_i^E of the Components in the Alloy Ti-1.85Ta-3.9W-2.05Mo-14.9Cr-70.3Ni-4.5Al (mass%) at 1500 K [1992Hil]. Reference states like in Table 4.

Alloy component i	a_i (at 1500 K)	μ_i^E (at 1500 K) [kJ·mol ⁻¹]
Cr	0.46 ± 0.09	12.9 ± 2.6
Ni	0.45 ± 0.11	-5.2 ± 2.9
Al	$(2.6 \pm 1.3) \cdot 10^{-6}$	-13.1 ± 8

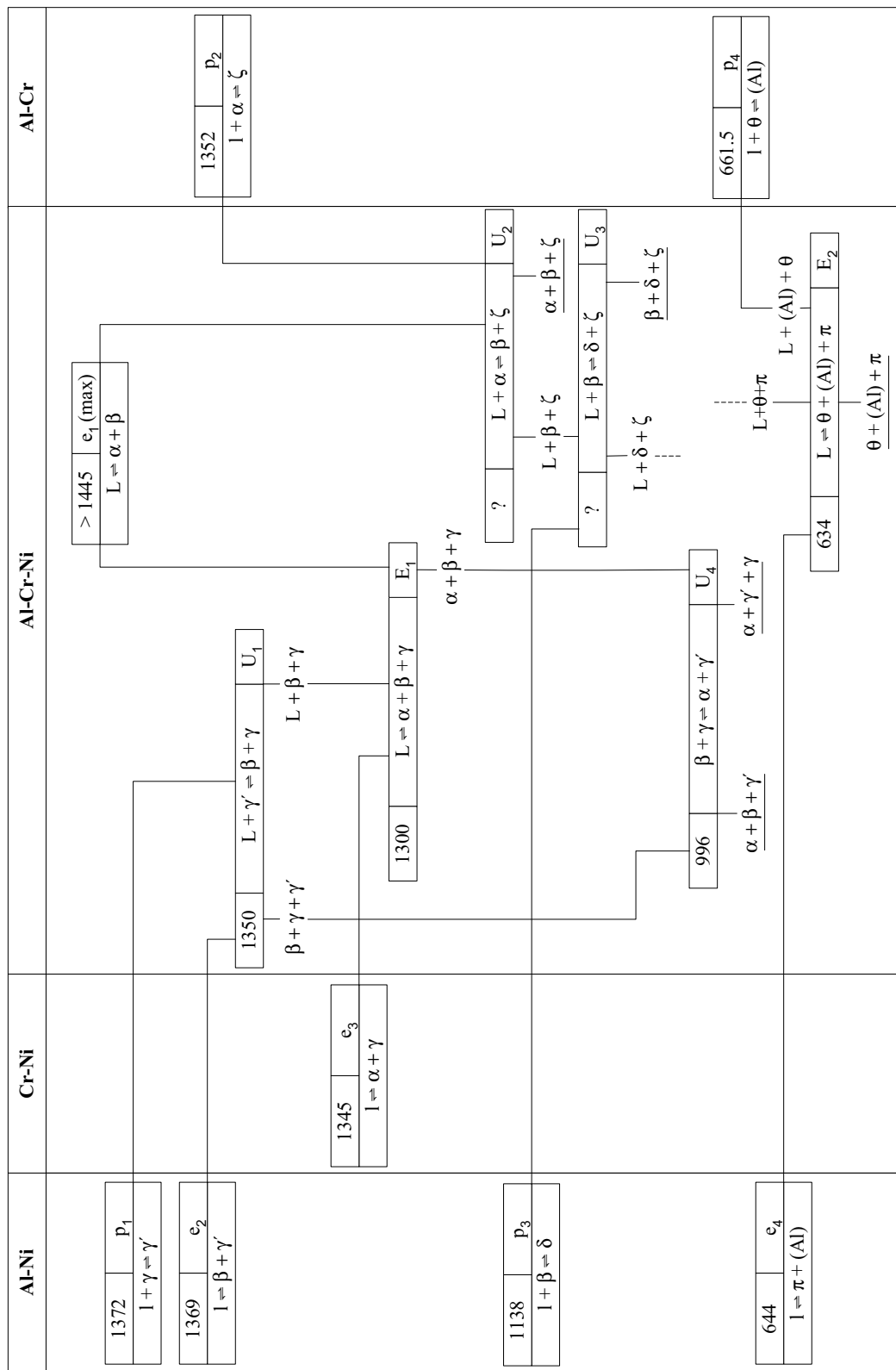


Fig. 1: Al-Cr-Ni. A partial reaction scheme

Fig. 2: Al-Cr-Ni.
Liquidus surface
projection at
Al < 65 at.%,
temperature in °C

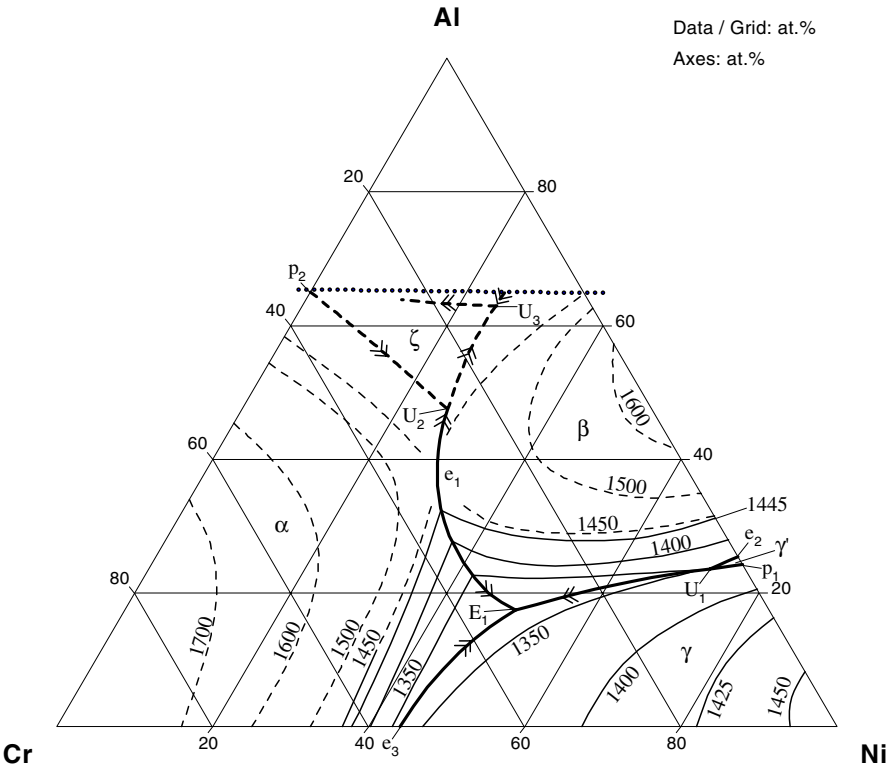


Fig. 3: Al-Cr-Ni.
 β -phase liquidus
surface in the vicinity
of $\text{Ni}_{0.5}\text{Al}_{0.5}$
composition after
[1993Cot1],
temperature in °C

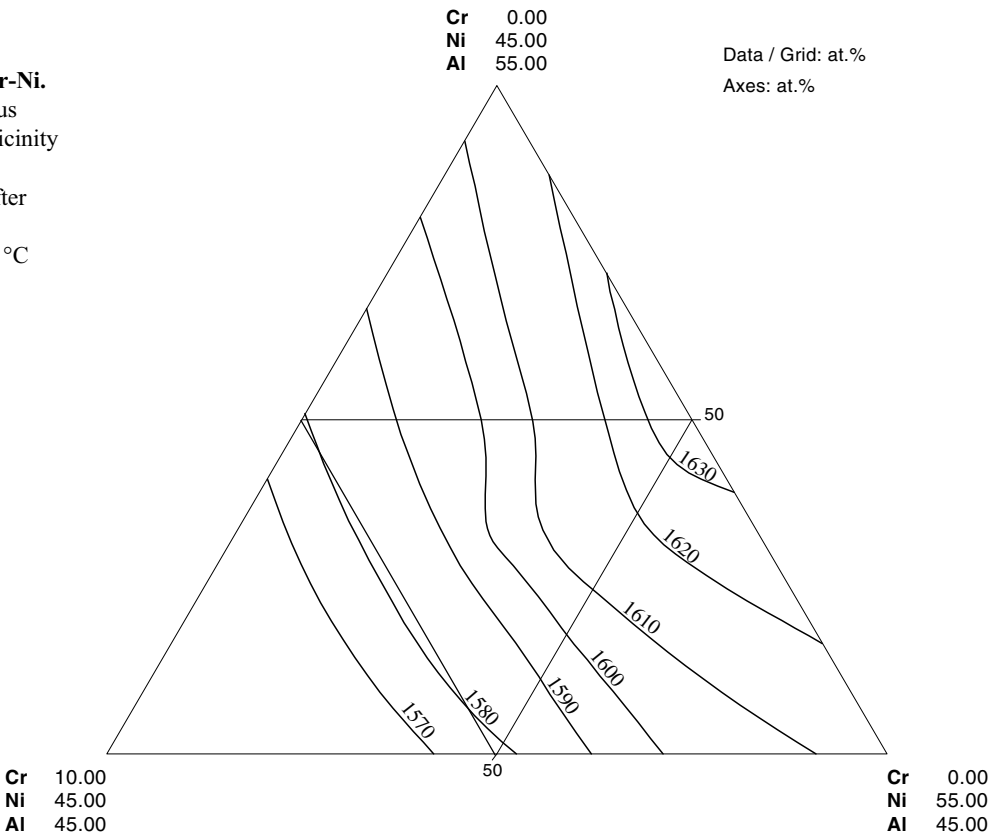


Fig. 4: Al-Cr-Ni.
Calculated partial
liquidus surface
projection
[2001Dup]
(existence of γ'
phase is not
considered),
temperature in $^{\circ}\text{C}$

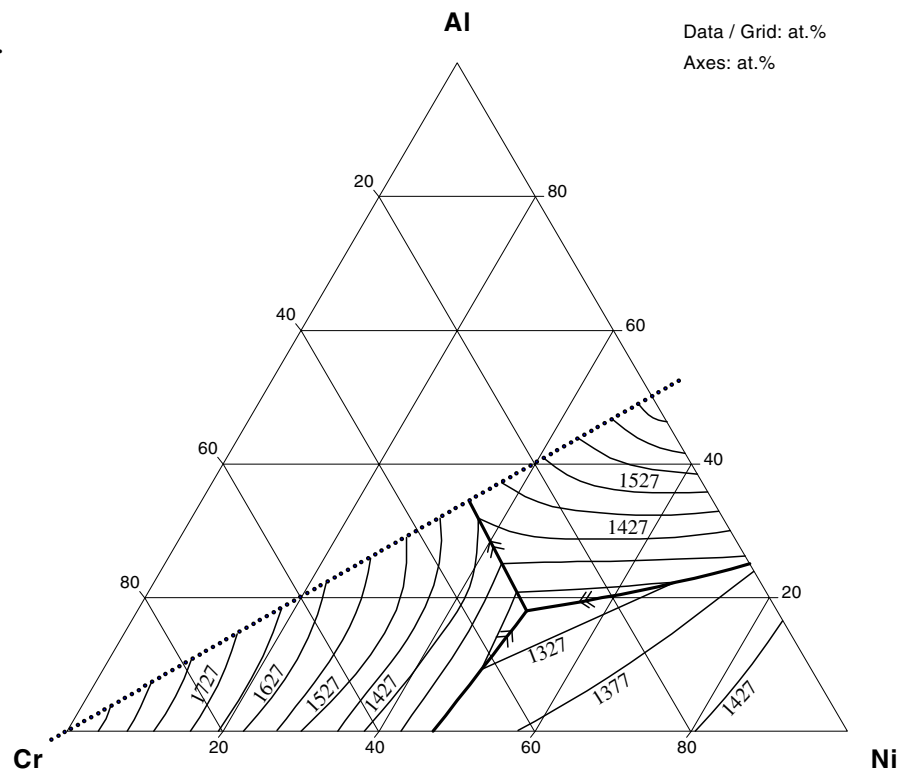


Fig. 5: Al-Cr-Ni.
Calculated partial
isothermal section at
1300 $^{\circ}\text{C}$ [2001Dup]

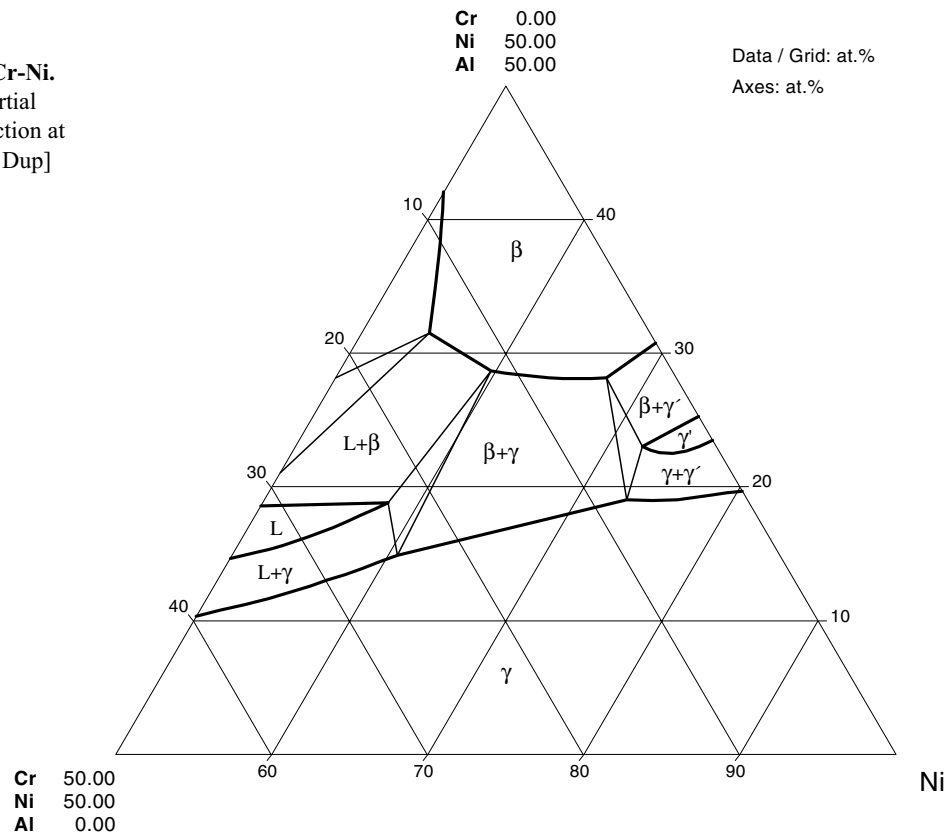


Fig. 7a: Al-Cr-Ni.
Partial isothermal
section at 1150°C
after [1985Ofo,
1991Rog]

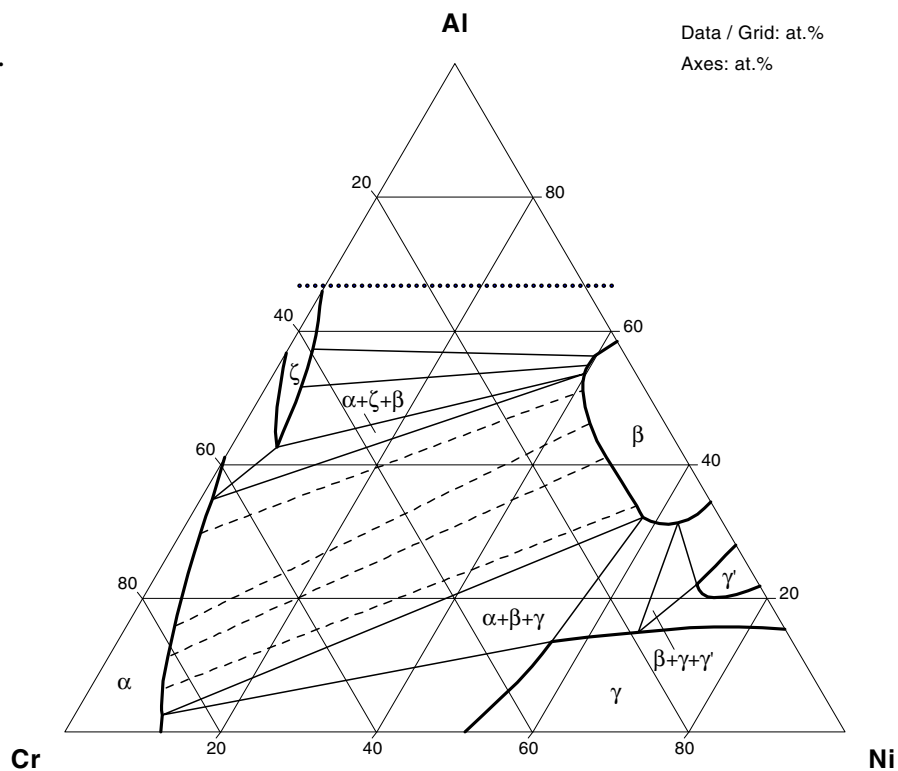


Fig. 7b: Al-Cr-Ni.
Partial isothermal
section at 1150°C:
solid lines are
calculated by
[2001Dup]; dashed
lines are experimental
data [1985Ofo]

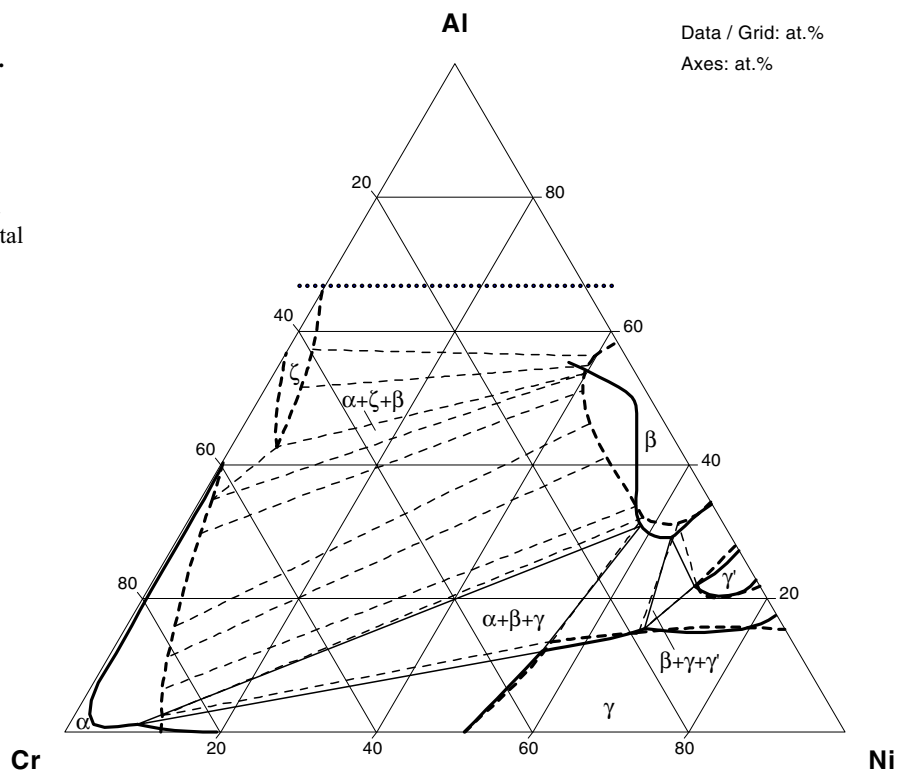


Fig. 8: Al-Cr-Ni.
Calculated partial
isothermal section at
1127°C [2001Dup]

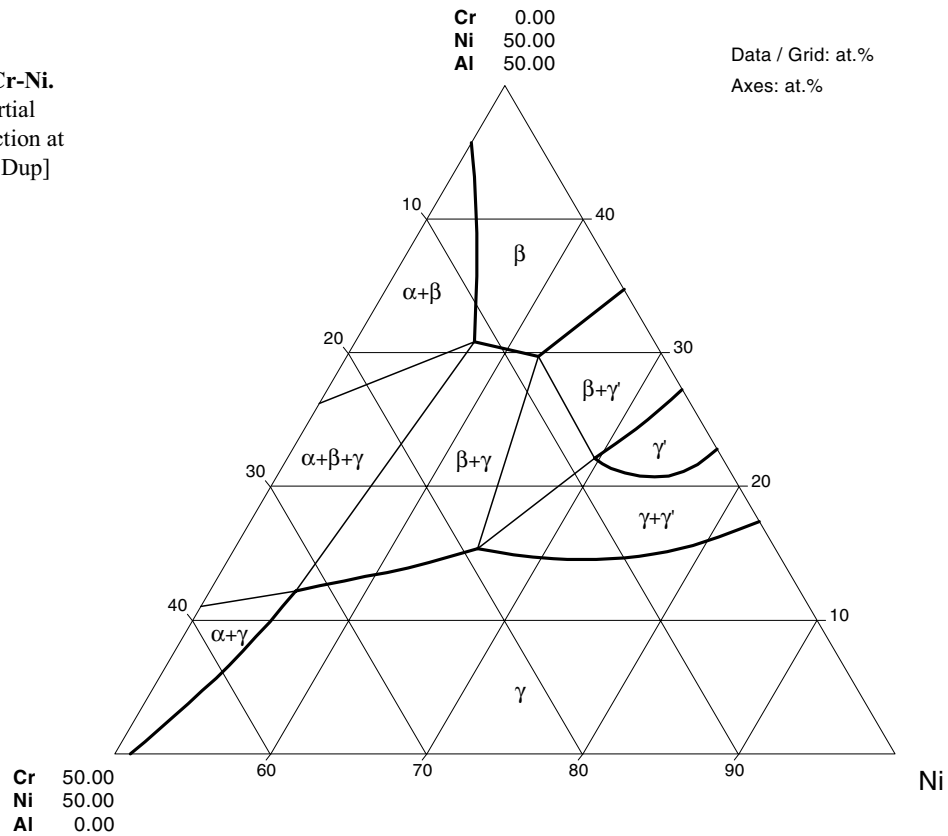


Fig. 9: Al-Cr-Ni.
Calculated partial
isothermal section at
1100°C [2001Dup]

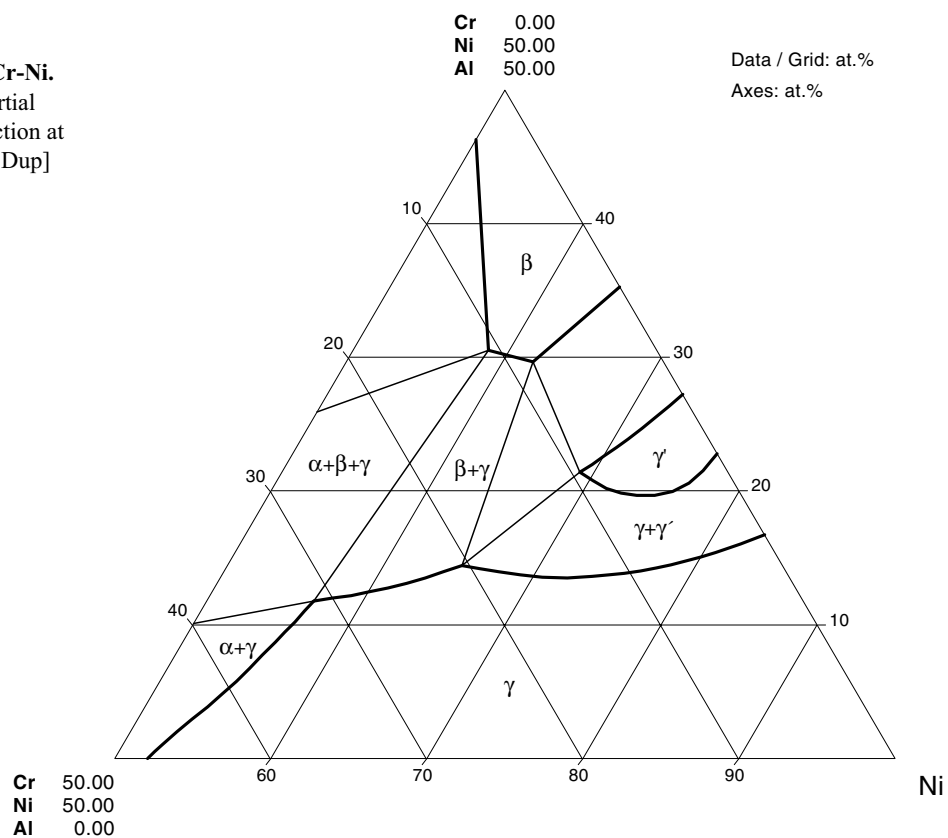


Fig. 10a: Al-Cr-Ni.
Partial isothermal
section at 1025°C
[1982Tu, 1984Mer]

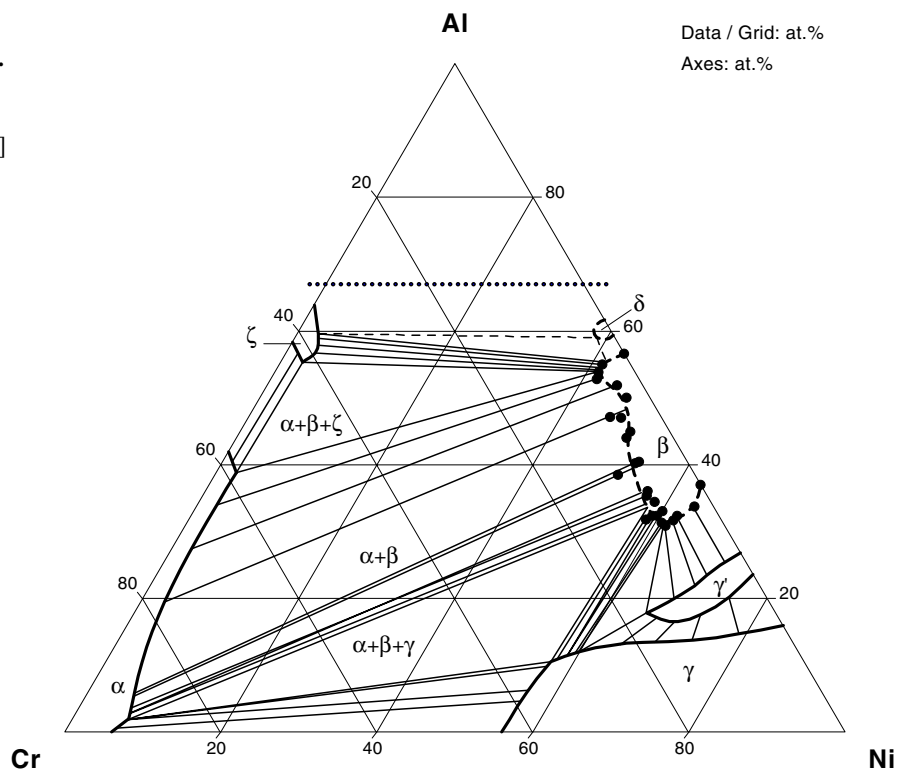


Fig. 10b: Al-Cr-Ni.
Partial isothermal
section at 1025°C
calculated by
[2001Dup]

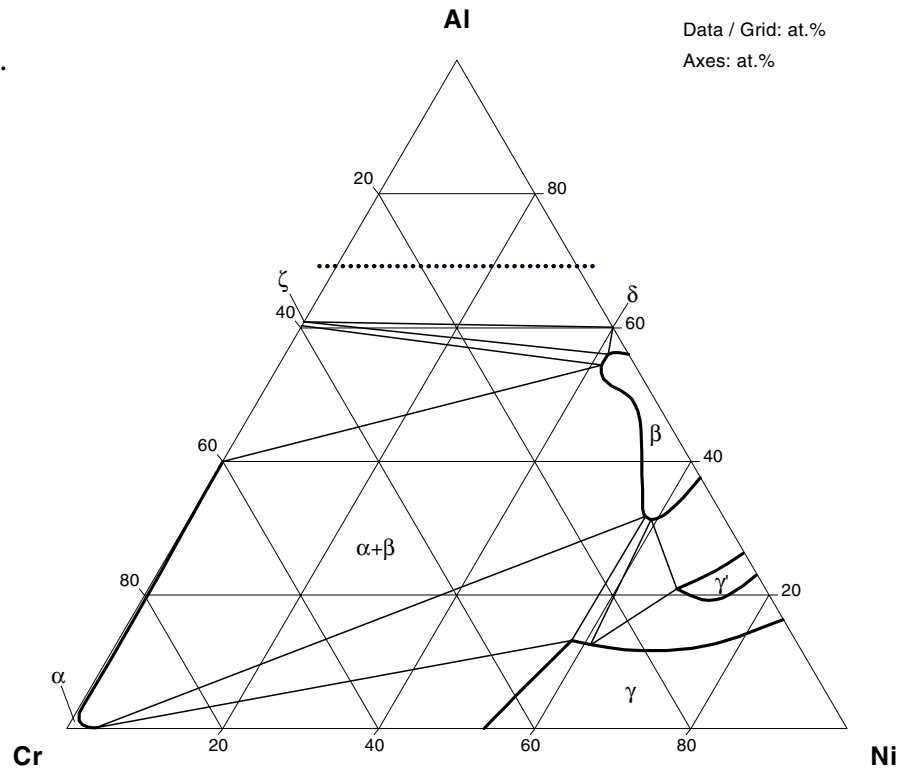


Fig. 11: Al-Cr-Ni.
Calculated partial
isothermal section at
1000°C [2001Dup]

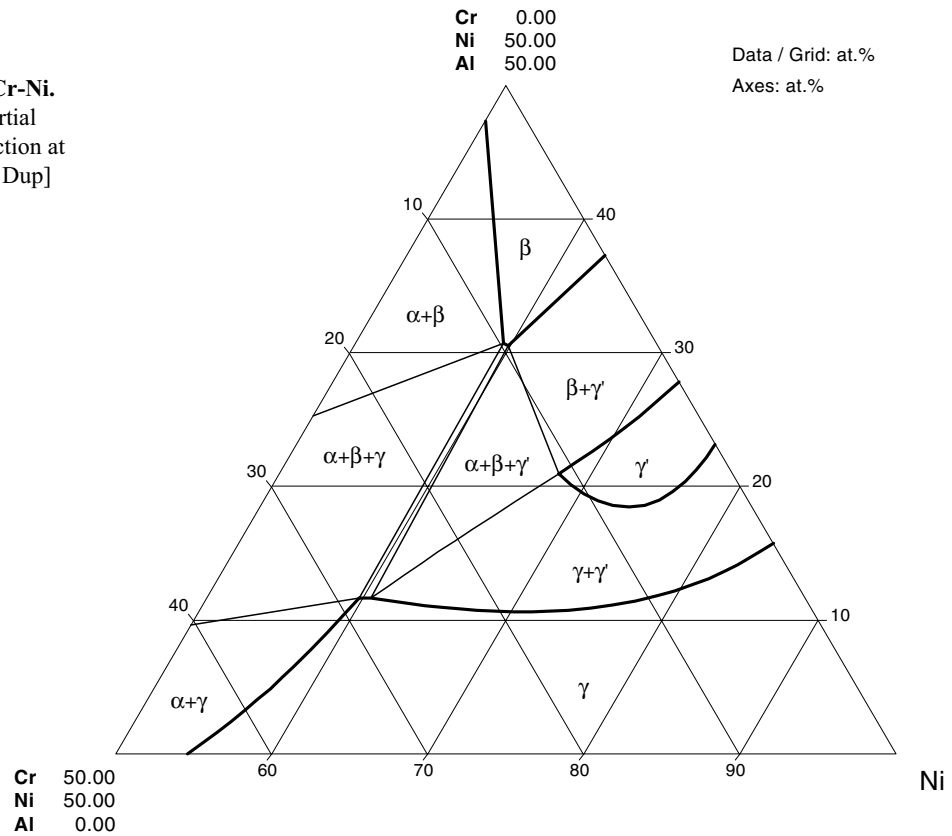


Fig. 12: Al-Cr-Ni.
Calculated partial
isothermal section at
900°C [2001Dup]

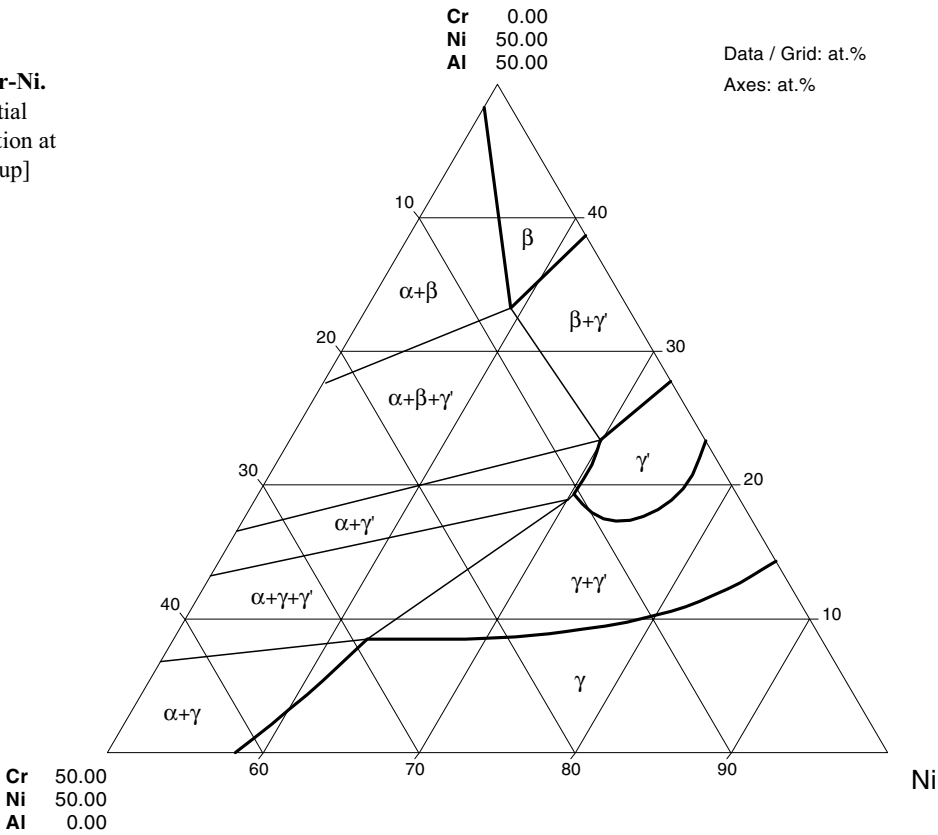


Fig. 13a: Al-Cr-Ni.
Partial isothermal
section at 850°C after
[1952Tay2, 1991Rog]

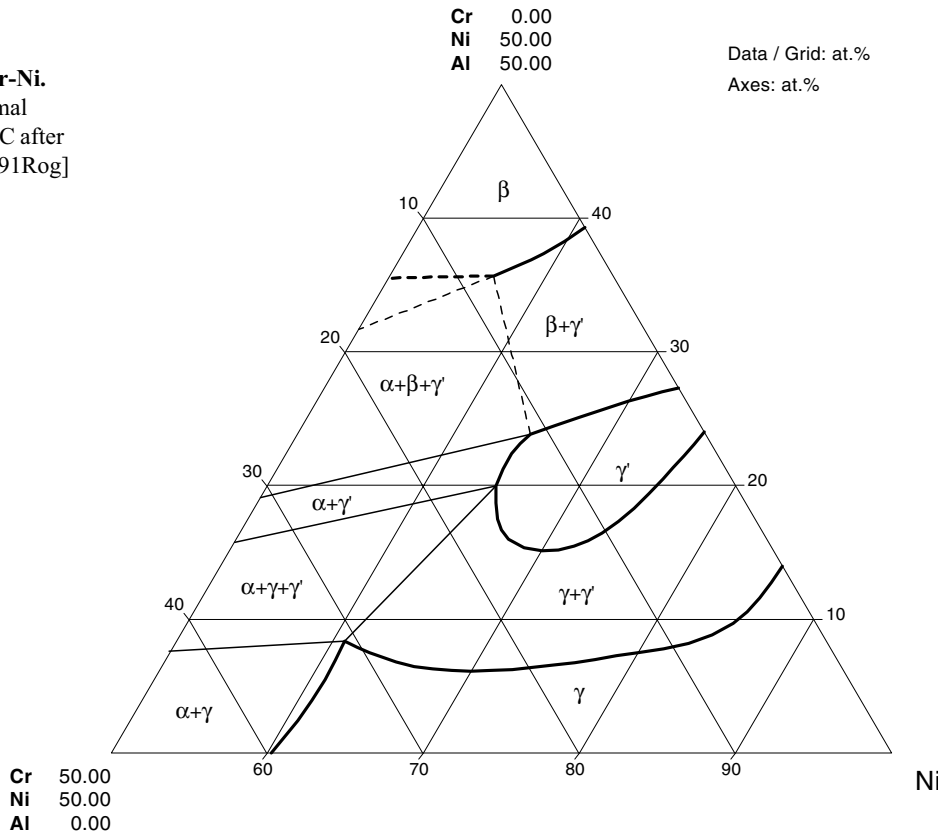


Fig. 13b: Al-Cr-Ni.
Partial isothermal
section at 850°C:
solid lines are
calculated by
[2001Dup]; dashed
lines are experimental
data [1952Tay2]

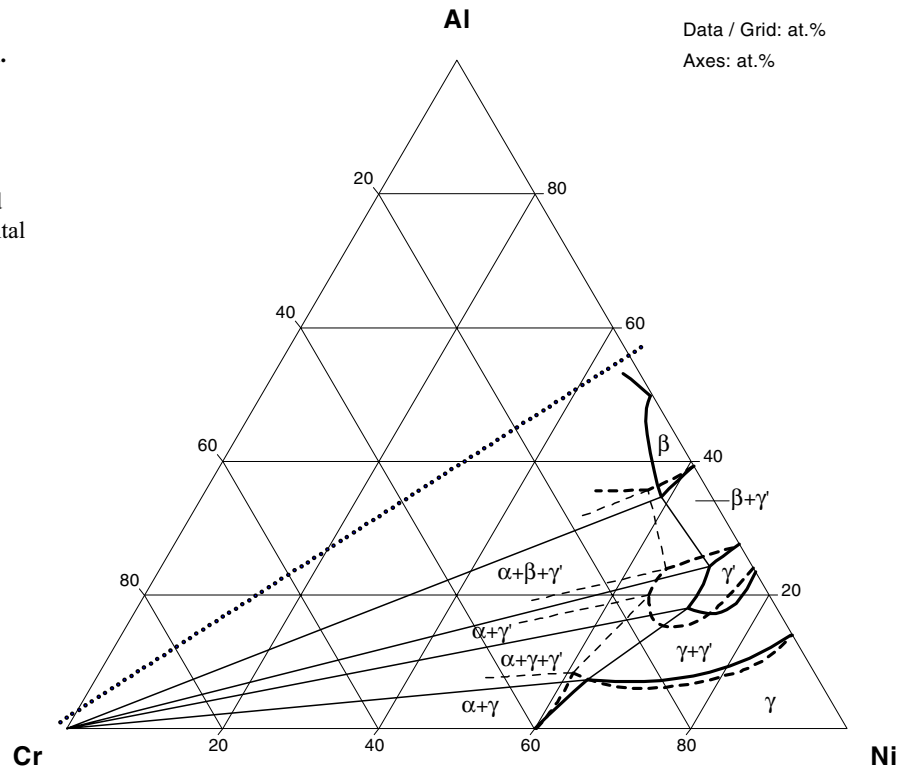


Fig. 14: Al-Cr-Ni.
Calculated partial
isothermal section at
800°C [2001Dup]

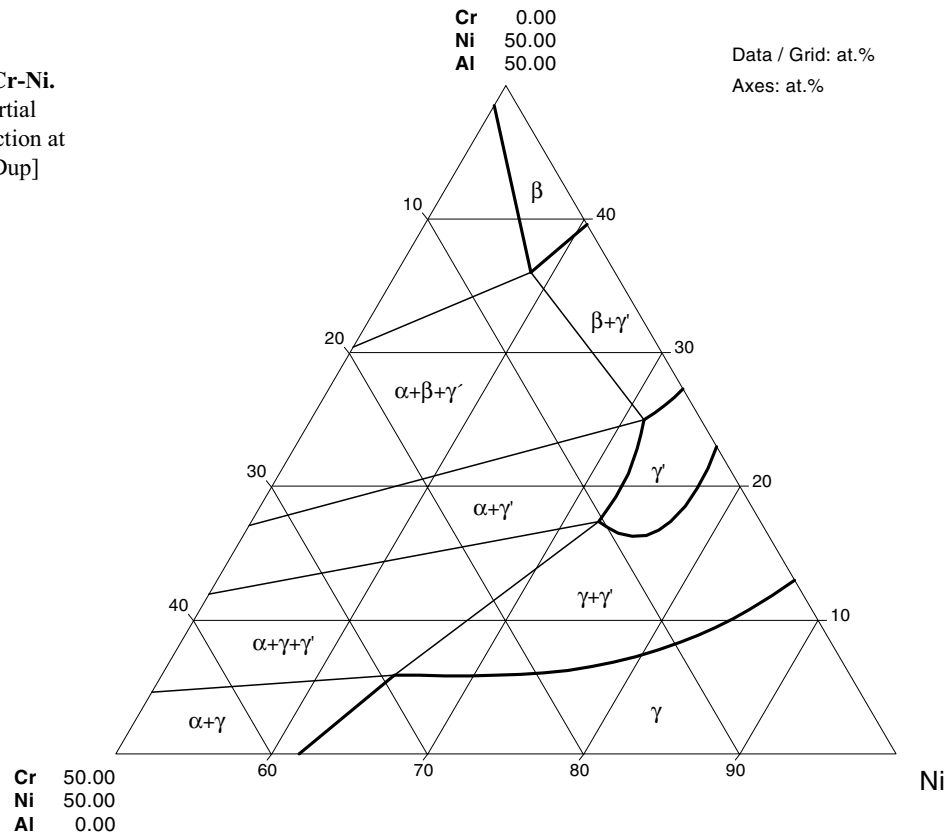


Fig. 15a: Al-Cr-Ni.
Partial isothermal
section at 750°C after
[1952Tay2, 1991Rog]

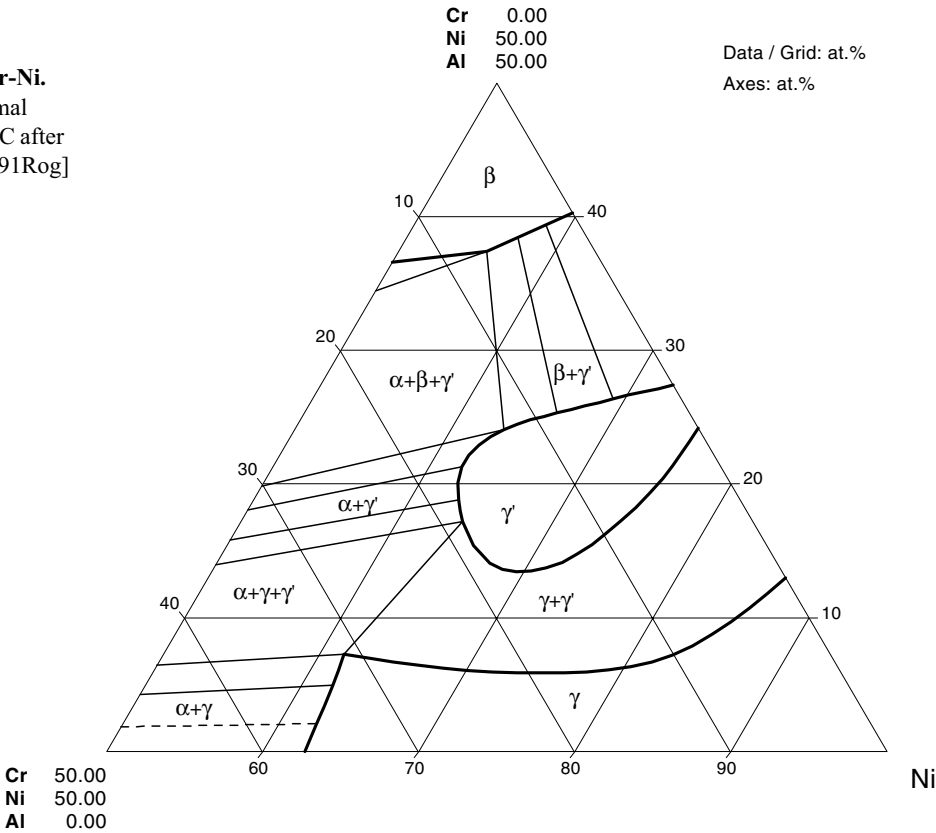


Fig. 15b: Al-Cr-Ni.
Partial isothermal
section at 750°C:
solid lines are
calculated by
[2001Dup]; dashed
lines are experimental
data [1952Tay2]

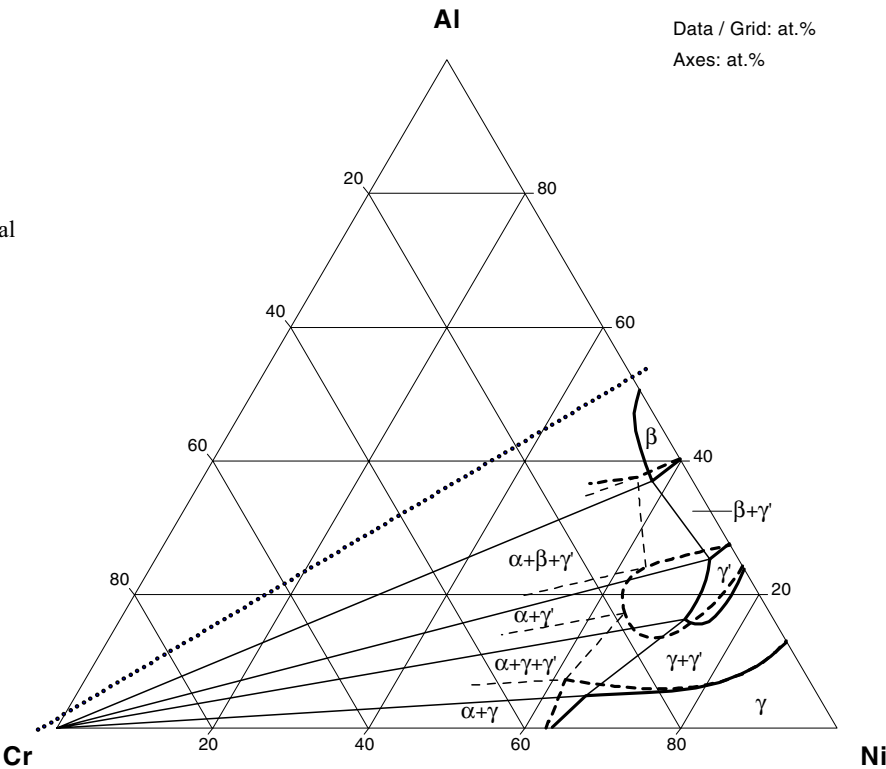


Fig. 16: Al-Cr-Ni.
Solvus isotherm of β -
phase at about 600°C
in the vicinity of
 $\text{Ni}_{0.5}\text{Al}_{0.5}$
composition after
[1993Cot1]

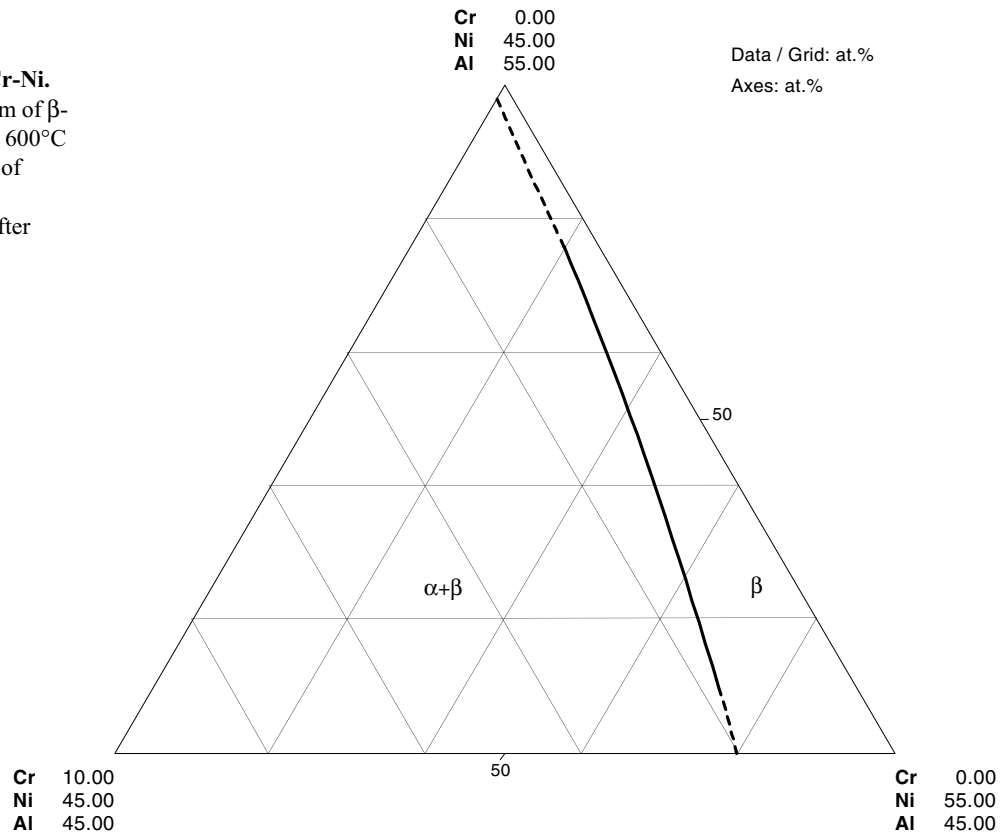


Fig. 17: Al-Cr-Ni.
 β -Phase solidus
surface in the vicinity
of $\text{Ni}_{0.5}\text{Al}_{0.5}$
composition after
[1993Cot1],
temperature in °C

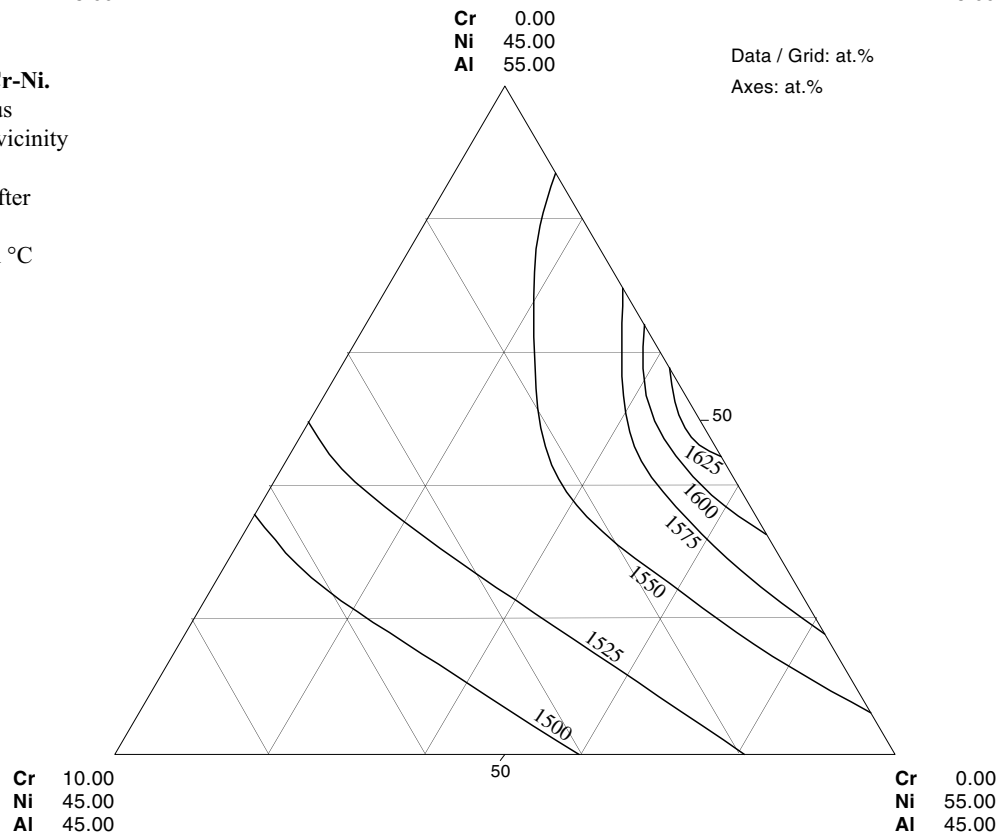


Fig. 18: Al-Cr-Ni.
Polythermal section
Cr - 50Ni50Al
[2001Dup]

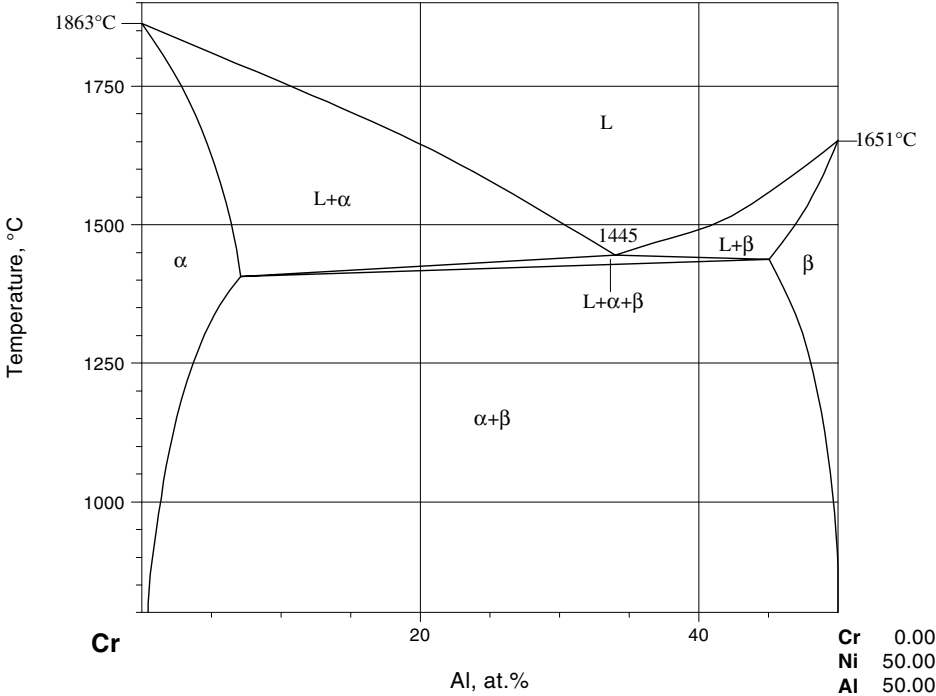


Fig. 19a: Al-Cr-Ni.
Polythermal section
Ni₃Cr - Ni₃Al, after
[1952Tay2, 1984Mer]

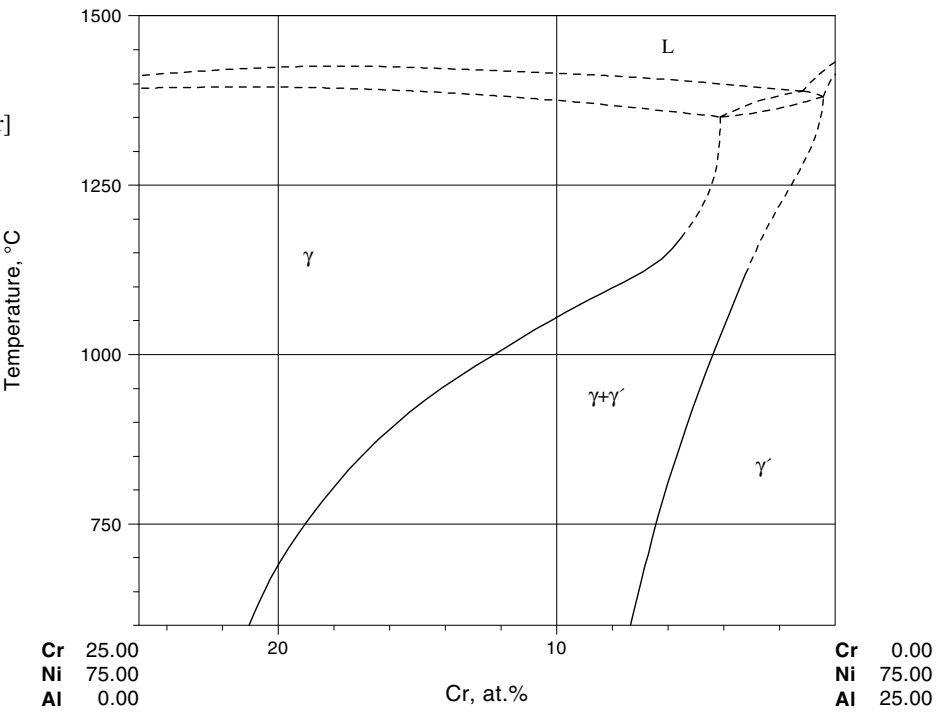


Fig. 19b: Al-Cr-Ni.
Calculated
polythermal section
Ni₃Cr - Ni₃Al
[2001Dup]

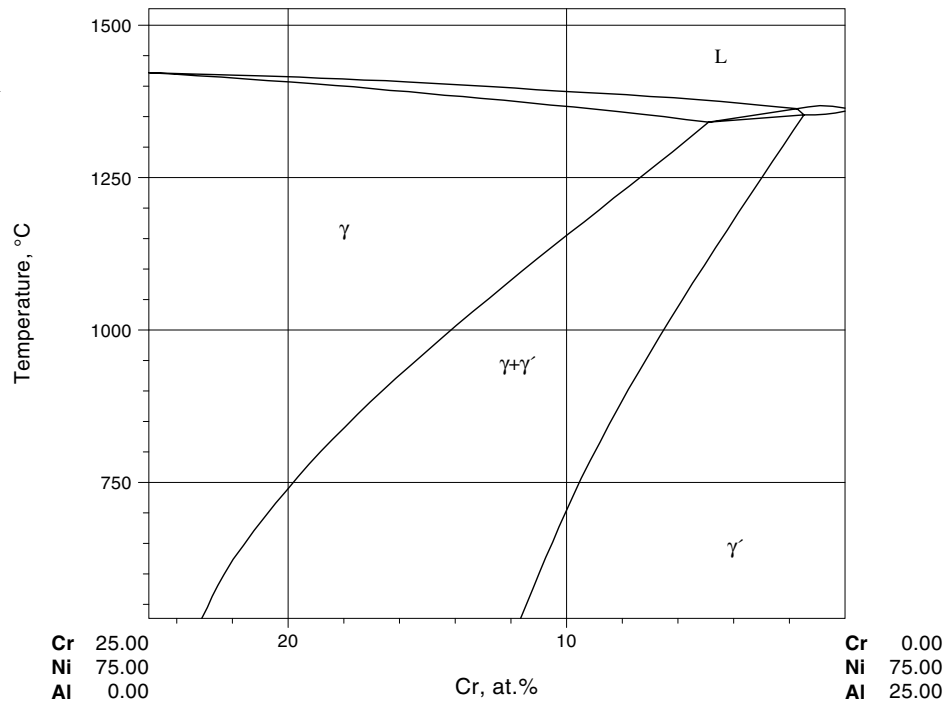


Fig. 20: Al-Cr-Ni.
Polythermal section
53Cr47Ni
-40Cr45Ni15Al
[1991Gor]

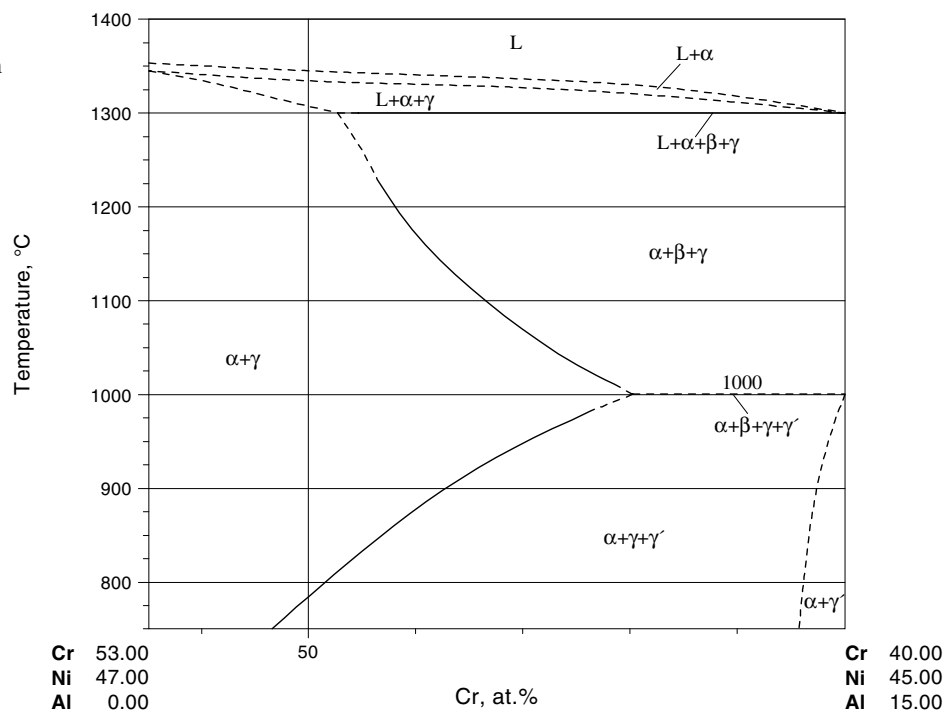


Fig. 21: Al-Cr-Ni.
Polythermal section
39Cr43Ni18Al
-34Cr33Ni33Al
[1992Gor]

