

phase, and its (010) rectangular cells are shown together with a simulated image in the bottom of Figure 1(a). If these faults occur repeatedly at every (100), then a new orthorhombic phase will result (Figure 1(b)). The oblique cells of $\text{Al}_{13}\text{Co}_4$ are outlined by white lines and their zigzag paths show the unit-cell faults (also marked with arrowheads in the upper part of this image), though some $\text{Al}_{13}\text{Co}_4$ still appears at the right of Figure 1(b). The rectangular cells thus formed are marked with black squares, and one of these cells is compared with the simulated image of Al_3Co . Again, the agreement is good. In fact, it has the *same* cell volume as the monoclinic $\text{Al}_{13}\text{Co}_4$, and it intergrows frequently with the latter, thanks to the common pentagonal and 36 deg rhombic subunits. Thus, the orthorhombic phase called Al_3Co in the literature can equally well be called the orthorhombic $\text{Al}_{13}\text{Co}_4$.

With regard to (3), it was shown in Reference 1 that the Al-Co decagonal quasicrystal transformed to the monoclinic $\tau^2\text{-Al}_{13}\text{Co}_4$ below 800 °C and that the reverse transformation occurred between 800 °C and 900 °C. Thus, it was concluded that the Al-Co decagonal quasicrystal was a stable phase within this temperature range. However, Drs. Grushko and Freiburg believed "no strong evidence in favor of this conclusion was presented" and they argued "to date, all quasicrystals classified as stable have been reported for ternary compositions." Furthermore, they maintained that the ternary Al-Co-Cu decagonal phase has a tendency to be stable at a higher temperature, with the decrease of the Cu content. More work should be done to prove this is true in very low Cu contents, or even with no Cu in the binary Al-Co decagonal phase.

It was pointed out^[2] that the τ^2 -inflated, monoclinic $\tau^2\text{-Al}_{13}\text{Co}_4$ phase is a better approximant to the Al-Co decagonal phase than $\text{Al}_{13}\text{Co}_4$. Now we have found a τ^3 -inflated $\tau^3\text{-Al}_{13}\text{Co}_4$ phase with even larger lattice parameters: $a = 6.4$, $b = 0.81$, and $c = 5.2$ nm, and $\beta \approx 108$ deg. This is an even better approximant. In Figure 2, the [010] EDP taken along the **b** axis of this rational approximant τ^3 is compared with the tenfold EDP of the decagonal quasicrystal (DQC) in a composite EDP. The 0 0 5, 0 0 8, 0 0 13, and 0 0 21 diffraction spots of the τ^3 phase coincide almost exactly with the strong spots of the DQC in τ -inflated ratios. Therefore, the spot rows of the DQC can be well extended into the τ^3 phase, and the pentagons marked with arrows and dots in these two phases appear similar as far as strong spots are concerned. From the approximant point of view, the decagonal quasicrystal can be considered as the end member of this series of τ -inflated phases with an infinitely large unit cell.

REFERENCES

1. X.L. Ma and K.H. Kuo: *Metall. Mater. Trans. A*, 1994, vol. 25A, pp. 47-56.
2. X.L. Ma and K.H. Kuo: *Metall. Trans. A*, 1992, vol. 23A, pp. 1121-28.

Improved Thermochemical Calculations for Maraging Steels and Comparison with Atom Probe Measurements of Phase Composition

W. SHA and G.D.W. SMITH

A series of studies of phase transformations in VASCOMAX C-300 and T-300* maraging steels (for

*VASCOMAX C-300 and T-300 are trademarks of Teledyne Vasco (formerly Vanadium Alloy Steel Company), Latrobe, PA.

composition, see Table I) has been published by the authors, based mainly on atom probe analysis and thermochemical calculations.^[1-4] It was experimentally observed that decomposition started with the formation of Mo-enriched Ni_3Ti phase in the martensite matrix and Fe_7Mo_6 phase formed at a later stage of aging. While generally good agreement was achieved between the experimental studies and the thermodynamical modeling, a significant drawback was that the Ni_3Ti phase, which was invariably present in these aged maraging steels, was not included in the calculations. This was a result of the lack of information in the available databases at the time the work was done. This note reports the results of more recent calculations using an updated database that contains all the phases relevant in these alloy systems.

Calculations were conducted with Thermo-Calc version H,^[5] with Kaufman database.^[6] Six phases were involved in the calculation: bcc, fcc, Ni_3Ti , Ni_3Mo , $(\text{Fe}, \text{Co})_2\text{Ti}$ Laves, and Fe_7Mo_6 μ . The equilibrium of C-300 at 510 °C consisted of bcc, fcc (41.2 mol pct), and μ (4.5 pct). When fcc was excluded in the calculation, Ni_3Ti (2.7 pct) also became thermodynamically stable. The driving forces for the formation of different phases in a bcc matrix are given in Table II. (Here, as in References 1 and 2, a positive $\Delta G/RT$ indicates stability of the phase involved. The units of ΔG are J mol^{-1} , and R is expressed in $\text{J mol}^{-1} \text{K}^{-1}$. The overall quantity ($\Delta G/RT$) is therefore dimensionless.)

The equilibrium of T-300 at 510 °C consisted of bcc, fcc (36.3 pct), Ni_3Ti (3.0 pct), and μ (4.0 pct). When fcc was excluded, the equilibrium became bcc, Ni_3Ti (8.6 pct), and μ (4.2 pct). The calculated driving forces in a bcc matrix are included in Table II.

These new and improved calculation results are in excellent agreement with the atom probe microanalytical work published previously. When applying the Thermo-Calc software to steel systems, it appears that the driving force information obtained gives a good measure of the tendency for precipitation.^[7] This is supported by the

W. SHA, formerly Graduate Student, Department of Materials, Oxford University, is Research Fellow, the Department of Materials Science and Metallurgy, Cambridge University, Cambridge CB2 3QZ, United Kingdom. G.D.W. SMITH, University Reader, is with the Department of Materials, Oxford University, Oxford OX1 3PH, United Kingdom.

Manuscript submitted August 10, 1993.

Table I. Nominal Composition of the C-300 and T-300 Maraging Steels (Atomic Percent)

	Ni	Co	Mo	Ti	Fe
C-300	18.2	8.8	2.9	0.7	balance
T-300	18.0	—	2.4	2.2	balance

Table II. Driving Forces ($\Delta G/RT$) for Precipitation of Phases in C-300 and T-300 Maraging Steel Systems at 510 °C: Thermo-Calc^[5] Thermochemical Calculation Results*

Phase	Driving Force in C-300	Driving Force in T-300
Ni ₃ Ti	0.95	1.09
Ni ₃ Mo	0.64	0.33
(Fe, Co) ₂ Ti Laves	-0.82	-0.24
Fe ₇ Mo ₆ μ	0.38	0.42
Fcc	0.74	0.67

*In our earlier calculations (Table X of Ref. 1), fcc austenite was included as a stable phase. The data presented here refer to the decomposition of a bcc matrix, with fcc austenite excluded, which is a more realistic model for the initial stage of decomposition of a maraging steel.

current calculation results and the earlier atom probe experimental analysis data, where Ni₃Ti has the highest driving force and is always the phase first to precipitate out from the bcc matrix. The failure to observe Ni₃Mo precipitation in C-300 steel is, at first, surprising. However, it should be noted from our earlier atom probe results^[1] that the initial clusters to form in C-300 were rich in both Ti and Mo. Naturally, kinetics must also play a part in the decomposition process. For example, the formation of austenite (fcc) is delayed until well after precipitation of intermetallics occurs in the case of C-300. In T-300, austenite does not form after very prolonged aging.

We consider that the combination of thermochemical calculation with atomic-scale measurement of phase composition now provides a strong basis for future alloy development programs.

The atom probe research in Oxford was supported by the Science and Engineering Research Council (UK) under Grant No. GR/F/06111. WS is grateful to The British Council for financial support. The authors would like to thank Professor G.B. Olson for helpful discussions.

REFERENCES

1. W. Sha, A. Cerezo, and G.D.W. Smith: *Metall. Trans. A*, 1993, vol. 24A, pp. 1221-32.
2. W. Sha, A. Cerezo, and G.D.W. Smith: *Metall. Trans. A*, 1993, vol. 24A, pp. 1233-39.
3. W. Sha, A. Cerezo, and G.D.W. Smith: *Surf. Sci.*, 1991, vol. 246, pp. 278-84.
4. W. Sha, A. Cerezo, and G.D.W. Smith: *J. Phys. Colloq.*, 1989, vol. 50-C8, pp. 407-12.
5. B. Sundman, B. Jansson, and J.-O. Andersson: Paper presented at *ASM Meeting*, Orlando, FL, 1986.

6. L. Kaufman: *CALPHAD*, 1984, vol. 8, pp. 25-66.

7. G.B. Olson: Northwestern University, Evanston, IL, private communication, 1992.

Load-Signature Analysis for Pack Rolling of Near-Gamma Titanium Aluminide Alloys

S.L. SEMIATIN and V. SEETHARAMAN

Titanium-aluminide alloys based on the ordered equi-atomic TiAl phase are receiving increasing attention as candidates for advanced aerospace applications.^[1] Of greatest interest are the so-called near-gamma alloys, which contain a small amount (typically 10 to 20 pct) of the ordered alpha-two, or Ti₃Al, phase. With proper control of the microstructure, such materials possess a good blend of elevated temperature strength and creep resistance and a modicum of ambient temperature ductility and fracture toughness.^[2,3] The most versatile means to control the microstructure in the near-gamma titanium-aluminide alloys is through thermomechanical processing of ingot-metallurgy products. Hot-working processes, such as conventional extrusion, isothermal forging, and conventional forging, have been successfully employed for ingot breakdown.^[4-7] Secondary processing routes have included isothermal closed-die forging, hot pack rolling, and isothermal rolling,^[8,9,10] through these techniques, subscale turbine engine components^[11] and flat sheets as large as 400 × 700 × 1.7 mm^[9] have been produced.

The objective of the present work was to demonstrate the sensitivity of rolling-load signature analysis as a means to monitor hot-pack-rolling processes for the fabrication of sheet of near-gamma titanium-aluminide alloys and other difficult-to-work materials. In previous work,^[12] a simple method was developed for the prediction of temperature transients during two steps in the pack-rolling process: the transfer of the pack from the furnace to the rolling mill and the actual rolling operation itself. The accuracy of the temperature-transient calculations was established through load-signature data obtained during trials for Ti-48Al (atomic percent) rolled at a single nominal (furnace) temperature. In the present work, additional results are presented and discussed for hot pack rolling at various nominal temperatures and for a range of near-gamma titanium-aluminide alloys.

Five ingot metallurgy, near-gamma titanium-aluminide alloys (Table I) were used in the present work. Following triple arc melting and hot isostatic pressing

S.L. SEMIATIN, Senior Scientist, is with the Metals and Ceramics Division, Wright Laboratory Materials Directorate, WL/MLLN, Wright-Patterson Air Force Base, OH 45433-7817. V. SEETHARAMAN, Senior Scientist, is with UES, Inc., Dayton, OH 45432.

Manuscript submitted February 1, 1994.