On the Modeling of Vacuum Arc Remelting Process in Titanium Alloys

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Abstract. Mathematical modeling is routinely used in the process development and production of advanced aerospace alloys to gain greater insight into the effect of process parameters on final properties. This article describes the application of a 2-D mathematical VAR model presented at previous LMPC meetings. The impact of process parameters on melt pool geometry, solidification behavior, fluid-flow and chemistry in a Ti-6Al-4V ingot is discussed. Model predictions are validated against published data from an industrial size ingot, and results of a parametric study on particle dissolution are also discussed.

1. Introduction

The basic process of VAR is the continuous melting of a consumable electrode by means of a direct current arc under vacuum on the order of 0.1 to 1 Pa. In titanium, the consumable electrode is manufactured using compacted sponge and master alloys or is the output of a previous melt cycle (either VAR or Cold Hearth melted) /1/. The consumable electrode is suspended in a water cooled copper crucible. A DC electric arc is struck between the bottom of the VAR crucible (stool or baseplate) and the bottom of the electrode. The electrode melts creating an ingot within the crucible /2/. For most titanium and titanium alloys, melting currents are very high, and an external, current carrying coil around the water jacket is used to contain the motion of the arc.

The mathematical model, SOLAR, (SOLidification during Arc Remelting) is routinely used at TIMET for new process development and process improvement efforts. The model was developed at the École des Mines, Nancy, and over the past decade, several papers have been published (including LMPC proceedings) describing the model and highlighting its capabilities. In short, the ingot growth during VAR is simulated using a two-dimensional, axi-symmetric representation. The model computes the current, temperature, turbulent flow, species distribution, and particle dissolution in the ingot by iteratively solving the appropriate governing equations with prescribed boundary conditions using the finite volume method /3/.

The primary focus of this paper is to present additional results of model validation by comparing model predictions to measured data already available in the literature /4/. The predicted pool profile compared well against reported measurements from a 0.914 m (36 inch), Ti-6Al-4V ingot using nominal thermo-physical properties /5/. Here, the effect of processing parameters on pool profile and ingot structure is presented. The likelihood of dissolution of exogenous inclusions is also discussed.
2. Results

As reported /6/, a 0.914 m dia. ingot was melted at a constant current of 33 kA and voltage of 44.7 V, which resulted in a melt rate of 0.45 kg/sec. A 2 mT axial stirring field, reversing every 60 seconds, was applied during this melt. These parameters, along with the nominal thermo-physical properties served as input to the model. These VAR conditions produce a very deep pool, reaching a depth of nearly 1.5 m at the center with very steep side walls. The effect of stirring field strength on pool shape is shown in Figure 1. As the field is increased from 0.5 mT to 2 mT, the azimuthal flow resulting from the interaction of the axial stirring field and radial current intensifies. This, in turn, results in higher velocities and a steeper pool in the top half of the ingot. Pool depth also increases as the stirring field is increased.

The flow within the pool is very transient, and fluctuates during the stirring cycle. At the start of the stirring cycle (t=0), the in-plane flow \((V_r, V_z)\) has one large re-circulating anti-clockwise loop. During the cycle, this flow field changes and can form several loops in different directions. Azimuthal flow, on the other hand, gradually increases throughout the cycle and is highest at the top of the ingot where there is good contact with the crucible wall. The predicted flow field at the beginning of the cycle is shown in Figure 2. The maximum azimuthal velocity is of the order of 0.4 m/sec, whereas the maximum axial velocity is 0.05/sec under these operating conditions.

Material flow and solute partitioning within the mushy zone of the melt pool results in macro-segregation, as shown in Figure 3 for a hypothetical electrode composition and melt profile /7/. During solidification, iron segregates to the liquid \((k_{Fe}<1)\), resulting in nearly 15% enrichment of iron towards the center of the ingot. Oxygen, on the other hand, segregates to the solid, \((k_O>1)\), and there is nearly 10% depletion of oxygen at the center of the ingot. The overall segregation in the ingot is quantified by calculating a segregation index, \(SI = \frac{1}{V} \int \left| \frac{c-c_0}{c_0} \right| dV \) for each solute element /8/. Results indicate that as stirring field intensity increases, the segregation index also increases.

The estimated temperature gradient in the ingot is shown in Figure 4. The gradient is high, on the order of 10 K/mm, close to the lateral surface and drops three orders of magnitude towards the center to nearly 0.01 K/mm. Ingot growth is in the range of 0.05 to 0.01 mm/sec. Referring to the solidification maps available in the literature /9/, would indicate a change in grain structure from columnar at the surface (higher temperature gradients) to equiaxed at center. This is consistent with observation from literature /4/, with the columnar-to-equiaxed transition occurring near 0.1 K/mm.

2.1. Inclusion Dissolution

The two common exogenous particles found in titanium alloys are low density, high nitrogen inclusions (LDI) and high density refractory metal inclusions (HDI). These inclusions are potentially removed by the diffusion of nitrogen or the refractory element in the molten metal pool. The capability to predict particle dissolution has recently been added to the SOLAR model /10, 11/, and a parametric study is presented in this section.

Figure 5 shows the motion of a 0.5 mm dia. particle with a density nearly 2.5 time that of molten titanium, inserted 1 cm from the centerline of the ingot. The projection of the particle on the r-z plane as it moves in the pool is shown in Fig 5. In this instance, the particle drops straight to the bottom of the pool. Similarly, particles of different initial diameter were inserted into the pool, and the model shows that, for the given operating conditions, particles with initial diameter of 0.5 mm and above survive and remain in the ingot, and only smaller particles, of the order of 0.25 mm diameter, fully dissolve.

Using mass balance at the particle-liquid interface, the dissolution time for a spherical particle can be approximated by the following relationship /12/: \(t = \frac{d^2}{D \left( \rho_p/\rho_{Ti} \right) \left( C_i \right)} \), where \(d\) is the initial particle diameter, \(D\) is the solute diffusivity, \(C_i\) is the initial solute concentration, and \(\rho_p\) and \(\rho_{Ti}\) are the density...
of the particle and molten titanium, respectively. The Sherwood number (sh#) represents the ratio of convective to diffusive mass transport. This relationship clearly shows that the dissolution time of an HDI is proportional to the square of the initial particle size and the particle density. The particle residence time within the pool is the ratio of the pool depth to settling velocity, and in order for the particle to fully dissolve, the dissolution time must, of course, be less than the residence time.

The trajectory of 0.5 mm diameter LDI with high nitrogen content and density slightly higher than the molten metal is shown in Figure 6. In this particular case, the LDI follows a tortuous path in the pool and dissolves. Similar to the HDI, the effect of different initial particle size was investigated. Results indicate that the diameter actually increases due to the formation of an outer beta phase at lower nitrogen contents. Under the given conditions, although the residence time is large, it appears that particles with initial diameter greater than 1 mm would survive, and only LDI’s with initial diameter of the order of 0.5 mm would dissolve in the pool. Using the mass transfer correlations in the literature with flow velocities between 0.05 and 0.5 m/sec, the nitrogen mass transfer (h_m) from the LDI to the bulk is in the range of 2x10^-4 m/sec to 6.5x10^-4 m/sec. Higher flow velocity and higher temperature promote LDI dissolution /13/.

3. Conclusions

4. The VAR of a production size ingot of Ti-6Al-4V was simulated using the SOLAR code. Results indicate that the pool profile is significantly affected by the externally applied magnetic field, as the field increases, the flow and overall segregation tendency within the ingot also increases. The model also accurately predicts trends in the temperature gradient within the ingot, and could be used as an indicator of the columnar to equiaxed transition in grain structure. The capability of the model to predict the dissolution of both HDI and LDI appears to be reasonable.

5. References

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Figure 1. Effect of stirring field strength of pool profile during the VAR of a 0.94 m diameter ingot at full power. (a) 0.5 mT, (b) 1 mT, (c) 2 mT. The stirring field was reversed every 60 sec.

Figure 2. Calculated flow field at the beginning of the stirring cycle during VAR, field of 2 mT. The in-plane velocity is shown on the right ($V_r, V_z$) and the azimuthal velocity, $V_\theta$, is shown on right. The dotted line represents the liquidus and solidus isotherms.
Figure 3. Computed ingot chemistry (0.94 m dia.) after VAR and starting electrode chemistry (0.762 m diameter). (a) iron distribution, (b) oxygen distribution.

Figure 4. Predicted temperature gradient and growth velocity during solidification of a 0.94 m dia. VAR ingot.
Figure 5: The trajectory of a 0.5 mm dia. HDI inserted at the center of the ingot (left), and HDI dissolution in the VAR pool (above).

Figure 6: The trajectory of a 0.5 mm dia. LDI inserted at the center of the ingot (left), and LDI dissolution in the VAR pool (above).